THERMAL CYCLING AND CREEP BEHAVIOUR OF ALLOY 617 BOILER COMPONENTS

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Abstract

Because of its proven beneficial material properties, Alloy 617 is the most promising material for the use in pressurised steam boiler components subject to cyclic loads and temperatures that are increasing due to an improved process efficiency. The result of this research project is the successful expression of the material properties of Alloy 617 into temperature-stress-time-dependent equations and the formulation of damage accumulation by all relevant damage mechanisms. All considerations and assumptions for the compilation of the equations have been discussed and concluded, which grants an unconditional use of these equations up to 800 °C. The accumulation of damage is maximised by the limits of the ASME design code. A material model is created in FE software program ANSYS based on the material property equations. Also, the damage accumulation equations and most design code's limits are built into this material model. Simulations have been run and the results are as expected and satisfactory. A Toolbox is constructed that provides a fast and accurate one-page overview of the calculated material properties, damage accumulation and fulfilment of the design code's requirements. For several reasons, analysis of design calculations and can easily be extended with other materials next to Alloy 617.

Foreword

This thesis has been written for the completion of the master's degree in Materials Science and Engineering. This work has been performed under the supervision of Dr. Ir. Michael Janssen, at Delft University of Technology.

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Table of Contents

Abbreviations and Symbols List	vii
Abbreviations	vii
Symbols	vii
1. Introduction	1
1.1 Research scope	1
1.1.1 Problem description	2
1.1.2 Approach to solving the problem	2
1.2 Heat Recovery Steam Generator (HRSG)	2
1.2.1 Next generation of materials for HRSG harps	3
1.3 Solar Boilers	4
1.4 Material selection	5
1.5 ASME Design Code and choice for ASME	6
1.6 Report Structure	7
2. Alloy 617	9
2.1 Properties of Alloy 617	9
2.1.1 Chemical composition alterations	9
2.1.2 Grain size influence	11
2.2 Plastic behaviour	12
2.2.1 Proportionality Limit	12
2.2.2 Yield Strength	13
2.2.3 Tensile Strength	15
2.2.4 Yield strength and tensile strength results combined	17
2.3 Creep behaviour	18
2.3.1 Determination of creep mechanism	19
2.3.2 Creep rate equation	19
2.3.2.2 Temperature dependency of creep equation parameters	20
2.3.3 Influence of γ' particles on creep rate	22
2.3.3.1 Stress influence	24
2.3.3.2 Consideration of Orowan threshold stress for determination of creep	25
2.3.4 Stress relaxation	
2.4 Time to rupture behaviour	
2.5 Fatigue	
2.5.1 Fatigue curve construction	
2.5.2 Fatigue curve for temperatures lower than 425 °C	
2.5.2.1 Fatigue data in the temperature range 425 – 538 °C	
2.5.3 Faligue curve for temperatures above 538 °C	
2.5.4 Cycle could methods	
2.7 Croop Estigue	
2.7 Creep-Faligue interaction	
2.7.1 Creep damage accumulation	
2.7.1.1 Creep damage accumulation	
2.7.1.2. rungue durinage decumantion	Δ1
2.7.2 Hold-time influence	41 Д1
2.7.3 Damage accumulation model	
2.7.3.1 Uncoupled visco-plastic model	
2.7.3.2 Creep-fatigue interaction part of the model	

2.8 Design code requirements, restrictions and limitations	.44
2.8.1 Comparison of the maximum allowable stresses	.44
2.8.1.1 Maximum allowable stress based on yield strength	.45
2.8.1.2 Creep stress to rupture	.45
2.8.1.3 Stress to obtain maximum allowable creep strain	.46
2.8.1.4 Comparison of adjusted results from the equations and ASME values	.46
2.8.2 Fatigue design curve and the application of conservative factors	.47
2.8.3 ASME restrictions on the accumulated creep-fatigue	.48
3. Toolbox	49
3.1 Maximum allowable stress	.49
3.1.1 Toolbox part for maximum allowable stress	.49
3.1.2 Limit for cyclic thermal stress and application in the Toolbox	.50
3.1.3 Degree of constraint	.50
3.2 Creep prediction	.50
3.2.1 Creep time to rupture determination	.50
3.2.2 Creep strain caused by primary stress	.51
3.2.3 Creep strain influenced by stress relaxation	.51
3.2.4 Maximum allowable creep strain	. 52
3.3 Fatigue prediction	. 53
3.3.1 Stress amplitude determination	.53
3.3.2 Fatigue for temperatures below 425 °C	.53
3.3.3 Fatigue for temperatures above 538 °C	. 53
3.3.4 Fatigue damage	.54
3.4 Creep-fatigue determination	.54
3.4.1 ASME requirement fulfilment	.54
3.5 Toolbox output	. 55
3.5.1 Comparison 1: primary stress versus the maximum allowable stress	.55
3.5.2 Comparison 2: maximum allowable thermal stress to avoid ratchetting	.55
3.5.3 Comparison 3: time to rupture is longer than the exposure time	. 55
3.5.4 Comparison 4: exceeding of the maximum allowable creep strain	.56
3.5.5 Comparison 5: maximum allowable fatigue damage accumulation	.56
3.5.6 Comparison 6: fulfilment of ASME requirement for creep-fatigue	.56
3.6 Functionality of the Toolbox	.57
3.6.1 Observations about Alloy 617	.57
	.57
4. ANSYS simulations	59
4.1 Test specimen for FE simulations	.59
4.2 Additional physical material properties required for FE calculations	.62
4.3 Materials Properties Council Model	.63
4.3.1 Check 1: stress exceeding the maximum allowable yield strength	.64
4.4 Creep mechanism incorporated into the material model	.65
4.4.1 Time to rupture	.65
4.4.2 Check 2: time to rupture exceeding the exposure time of the component	.66
4.4.3 Creep strain	.6/
4.4.4 Check 3: creep strain exceeding the maximum allowable creep strain	.69
4.5 Faligue mechanism incorporated into the material model	. 70
4.5.1 ratigue at lower temperatures (Delow 425 °C)	.70
4.3.2 CHECK 4. THE HUMBER OF CYCLES TO Idilute (DEIOW 425 $^{\circ}$ C)	./⊥ 71
4.5.5 Latigue at higher temperatures (above 550 C)	./⊥ 77
דד טופטר א נווב וועוושבו טו טיטובא נט ומווערב (מטטיב גאס ט)	. / Z

5. Validation of the Toolbox	73
5.1 True stress – true strain curves	73
5.2 Maximum allowable stress based on the yield strength	75
5.3 Time to rupture	75
5.4 Creep strain	76
5.5 Fatigue for temperatures lower than 425 °C	78
5.6 Fatigue for temperatures higher than 538 °C	79
6. Discussions	81
6.1 Creep data from creep strain rate	81
6.2 Fatigue data	82
6.3 Creep-fatigue data	83
6.4 Strain-rate sensitivity	86
6.5 Dynamic strain aging	88
6.6 Pressurised tubes versus conventional tests	88
6.7 Material long-term exposure	90
6.8 Other Materials	90
7. Conclusions	91
References	

Appendices

Appendix A	– Creep	A1 – A10
Appendix B	- High temperature fatigue	B1 – B12
Appendix C	- Fitness-For-Service Assessment Procedure	C1 – C20
Appendix D	- Mathematics and explanations	D1 – D8
Appendix E	– Bree diagram	E1 – E26
Appendix F	- ASME Design Code	F1 – F10
Appendix G	- Method for testing creep-fatigue interaction	G1–G10

Abbreviations and Symbols List

Abbreviations

API	American Petroleum Institute
ANL	Argonne National Laboratory
ASME	American Society of Mechanical Engineers
ASTM	American Society for Testing and Materials
BSU	Boise State University
CF	Creep Fatigue
CMS	Controlled Material Specification
E-P	Elastic-Plastic
FCC	Face Centred Cubic
FE	Finite Element
FFS	Fitness-For-Service
HCF	High-cycle Fatigue
HRSG	Heat Recovery Steam Generator
INL	Idaho National Laboratory
LCF	Low-cycle Fatigue
LMP	Larson Miller Parameter
MPC	Materials Properties Council
NA	Nickel-based Alloy
NEM	Nederlandsche Electrolasch Maatschappij
ORNL	Oak Ridge National Laboratory
SD	Shakedown

Symbols

<i>A</i> _f Adjustment constant for the best-fit fatigue curve	
<i>A_{int.}</i> Interaction coefficient between creep and fatigue me	echanisms
<i>A</i> ₁ Fitting constant for stress-strain curve of micro-strain	n in MPC model
A ₂ Fitting constant for stress-strain curve of macro-strai	n in MPC model
<i>A_R</i> Constant in the Larson-Miller Parameter equation for	r creep rupture
$a_{1,2,3,4}$ Parameter for yield and tensile strength equations	
a_R Time dependent material constant for rupture stress	determination
lpha Coefficient of thermal expansion	
α_m Mean coefficient of thermal expansion (at temperatu	re between T_1 and T_2)
α^{or} Constant used in the Orowan stress to account for di	slocation core radius
<i>B_c</i> Temperature-dependent constant for creep equation	1
<i>B_f</i> Endurance limit of the best-fit fatigue curve	
B_R Constant in the exponential part of LMP equation for	creep rupture
b Burger's vector	
<i>b</i> _f Fatigue strength exponent (Basquin equation)	
$b_{1,2,3,4}$ Parameter for yield and tensile strength equations	
β Constant in regression analysis in stress-strain hyster	esis loop for creep-fatigue test
Constant in regression analysis in stress-strain hyster	esis loop for creep-fatigue test
C _K Kinetic Constant	
C _P Specific heat	
<i>C</i> _{srs} Material constant for strain rate sensitivity equation	
C_T Taylor factor (value of 3.1 for FCC structure)	

C_1, C_2, C_3	Input parameters for Norton's creep equation for ANSYS
C_{f}	Fatigue ductility exponent (Coffin-Manson)
<i>C</i> _{1,2,4}	Parameter for yield and tensile strength equations
D	Diffusivity or Diffusivity coefficient at temperature
D_c	Damage parameter for creep
D_f	Damage parameter for fatigue
D _{f,total}	Summation of damage parameters for fatigue from the different cycle types
D _{tot}	Total damage parameter
D_0	Maximum diffusion coefficient dependent on temperature
d_p	Diameter of precipitate
Ε	Elastic or Young's modulus
E _C	Compressive elastic modulus.
E_T	Tensile elastic modulus.
E_{yf}	Young's modulus of the fatigue curve being utilised
E_{ym}	Young's modulus at mean temperature (usually temperature of assessment)
ε_a	Strain amplitude
E _{a,e}	Elastic strain amplitude
$\mathcal{E}_{a,p}$	Plastic strain amplitude
$\varepsilon_{c,allow}$.	Allowable accumulated creep strain
$\mathcal{E}_{c,p}$	Creep strain evolved due to primary stress
$\mathcal{E}_{C,S}$	Creep strain evolved due to secondary stress
\mathcal{E}_{es}	Engineering stress
$\varepsilon_{inel.}(t)$	Creep strain accumulated due to stress relaxation at time (t)
ε _p	Offset strain for proportionality limit
\mathcal{E}_t	Strain at transition point (fatigue)
\mathcal{E}_{ts}	True strain
\mathcal{E}_{v}	Strain at yield point (at 0.2% yield strength)
Ê	Elastic strain at t=0 before start of relaxation
ε_1	True plastic strain in the micro-strain region of the stress-strain curve
\mathcal{E}_2	True plastic strain in the macro-strain region of the stress-strain curve
ε_{f}^{\prime}	Fatigue ductility coefficient
ė _c	Steady-state creep strain rate
$\dot{\varepsilon}_{c.min}$	Minimum steady-state creep strain rate
Ė _{c.1}	Creep strain rate in graph/figure at point 1
$\dot{\varepsilon}_{c,2}$	Creep strain rate in graph/figure at point 2
ė _{test}	Strain rate for tensile test simulation
$\Delta \varepsilon_a$	Change in strain amplitude (ε_a - ε_t)
$\Delta \varepsilon_e$	Change in elastic strain amplitude
$\Delta \varepsilon_{in}$	Inelastic strain range (sum of plastic and creep strains in fatigue cycle)
$\Delta \varepsilon_n$	Change in plastic strain amplitude
$\Delta \varepsilon_t$	Total strain range (in fatigue cycle)
f_v	Volume fraction of precipitates
γ	Nickel-based matrix
γ'	Gamma prime particles: Ni ₃ (Al,X) in NA with X = Ti, Nb, Ta
γ_1	Strain in micro-strain region
γ_2	Strain in macro-strain region
H	Prager-Drucker factor
I	Interface region
Κ	Model parameter in MPC model based on ratio of yield strength to tensile strength
Ko	Parameter dependent on crystal structure

k th	Cycle during fatigue with specific characteristic "k"
k_B	Boltzman's constant
k _c	Thermal conductivity
k _{cf}	Constant in regression analysis in stress-strain hysteresis loop for creep-fatigue test
к	Relaxation parameter
L_S	Edge-to-edge spacing between precipitates
Μ	Total number of cycles
m _{srs}	Strain rate sensitivity exponent
m_1	Curve fitting exponent in the MPC model, hardening coefficient in large strain region
m_2	Curve fitting exponent in the MPC model equal to true strain at tensile strength
μ	Shear modulus
N _f	Number of full cycles to failure
$2N_f$	Number of load reversals to failure
N _{f,ASME}	Lowest number of cycles to failure based on both conservative factors
$N_{f,ASME,N}$	Number of cycles to failure including the conservative factor on cycles to failure
$N_{f,ASME,\sigma}$	Number of cycles to failure including the conservative factor on stress amplitude
$N_f(2\Delta\sigma)$	Number of cycles to failure based on two times the stress amplitude
$N_{f,k}$	Permissible number of cycles (cycles to failure) of the type of the k th cycle
$2N_t$	Number of load reversals at transition point
^m N	Number of cycles at point 'm'
^{n}N	Number of cycles at point 'n'
$\Delta 2N$	Change in number of load reversals (in comparison to the transition point)
n	Exponent in creep equation for σ/μ ratio, which is temperature-dependent (par. 2.3)
n_f	Exponent in fatigue model
n_R	Negative number equal to slope of $\Delta \log \tau_r$ divided by $\Delta \log \sigma$ (par. 2.5)
n_k	Actual number of repetitions of the k th cycle
n _r	temperature-dependent material exponent for rupture stress determination
n _t	Number of time increments
n_1	First time increment
ν	Poisson's ratio
v_{cf}	Frequency of cycle in creep-tatigue test
P	Proportionality limit
Q_c	Activation energy for creep
ΔQ	Secondary stress range
$\Delta Q_{max.}$	Cas constant
	Bas constant
$R_y(T)$	Polynomial trend-line for temperature dependent yield strength
$R_T(I)$	Polynomial trend-line for temperature-dependent tensile strength
κ _σ S	Average stress to obtain a creen rate of 0.01%/1.000b (i.e. 10^{-7})
S _c	Average stress to cause runture at the end of 100 000 hour
$S_{R,avg}$	Minimum stress to cause runture at the end of 100,000 hour
$S_{R,min}$	limit for the secondary equivalent stress range
S _U	Maximum allowable stress at Trave
S _T max	Tensile strength at room temperature
$S_T(T)$	Temperature-dependent tensile strength
$S_{\rm N}$	Yield strength at room temperature
$S_{n}(T)$	Temperature-dependent vield strength
ΔS_{mk}	Equivalent stress range of combined primary and secondary stresses of k th cycle
$mn \Delta S_{max} = 2$	Von Mises equivalent stress range
runge	

σ	Stress (general) or sigma phase, a possible phase in Alloy 617
σ_a	Stress amplitude
$\sigma_{allow.}$	Maximum allowable stress
$\sigma_{YS,allow}$	Maximum allowable stress based on yield strength
σ_c	Stress for creep determination
$\sigma_{c,allow}$	Maximum allowable stress based on creep strain rate
$\sigma_{c,min}$	Onset stress for creep
σ_{es}	Engineering stress
σ_{fl}	Flow stress
σ_i	Initial stress
σ_{ij}	Stress component with i,j=1,2,3
σ'_{f}	Fatigue strength coefficient
$^{mn}\sigma_{ij}$	Stress component range between point 'm' and 'n' with i,j=1,2,3
$m \sigma_{ij}$	Stress component at point 'm' with i,j=1,2,3
$n \sigma_{ij}$	Stress component at point 'n' with i,j=1,2,3
σ_{max}	Maximum stress in the cycle
σ_{min}	Minimum stress in the cycle
σ_p	Primary membrane stress
$\sigma_{R,allow.}$	Maximum allowable stress based on the creep rupture strength
$\sigma_{R,avg.}$	Average creep rupture strength
$\sigma_{R,min.}$	Minimum creep rupture strength
σ_t	True stress
σ_{UTS}	Ultimate tensile strength
σ_y	Yield strength
$\sigma_{y,Tmax}$	Specified minimum yield strength at T_{max}
σ_0	Threshold stress
σ_1 , σ_2	Stress in graph/plot at resp. point 1 and point 2
$\Delta \sigma$	Stress reduction
$\Delta \sigma_{constraint}$	Stress developed due to full mechanical constraint of the component
$\Delta \sigma_{ij}$	Stress range component with i,j=1,2,3
T_{\perp}	Temperature in °C
T_h	Homologous temperature (T/T_{melt})
T_K	Temperature in Kelvin
T _{melt}	Melting temperature
	Temperature at intersection of two data trend-lines
I_1, I_2	Fixed temperatures (absolute), conveniently chosen
ΔI	Time or experience between inner and outer surface of a pressure part
m_{+}	Time of exposure time
n_{+}	Time at point in
ι t_	Time till runture
c_R	Minimum time till runture
Λt	Time segment
<u>т</u>	Shear stress
T _{climb}	Shear stress required for a dislocation to climb around a precipitate
$\tau_{ m eff}$	Effective shear stress
τ_{0r}	Orowan bowing stress
X	Ratio primary membrane stress to specified yield strength at T_{max}
Ζ	Zener-Hollomon parameter
	-

1. Introduction

In order to meet the increasing need of energy, designers work on improving the efficiency of power plants, research new methods to generate electricity and try to reduce the cost of the energy production. Adherent, designers develop sustainable energy solutions to meet environmental quota (e.g. CO₂ reduction), improve life expectation of the power plants and guarantee safe use. One of the energy production methods, which is conventional, robust and 'green', is the steam power plant. Commonly, the steam power plant is heated by fossil exhaust gasses, but recent actions are taken to use other means of heating, like solar radiation, in order to reduce the CO₂ emission.

Pressurised components are critical parts in a steam power plant. The design of these components has to meet the requirements of both design code and material standards. Due to design evolution and the development of new materials, these design codes and standards often lack material data. This lack is frequently compensated for by using correlations between data known, proven equations for material parameters and mechanisms expected to occur. Conservative factors are needed to ensure a safe design and to compensate for the lack of material data in the high temperature regime.

1.1 Research scope

The research is performed to investigate the material properties, the damage mechanisms active, the design methods and the limits of the design code for the use of Alloy 617 in pressurised components subject to thermal cycling at temperatures in the creep regime. The flow chart in figure 1.1 provides an overview of the research aspects, which cover other topics besides material scientific topics as well.



Figure 1.1: Flow chart of the research of applying Alloy 617 in pressurised components.

The research described in this report focusses especially on the material science part as shown in the right column of the flow chart in figure 1.1. As well, design code requirements and limits are considered as illustrated in the left column of the flow chart in figure 1.1. This is needed, because the design and the use of Alloy 617 in that design have to be evaluated on reliable, ensured and safe application. Both material science and design code aspects are considered in the design of the pressurised components, which is covered in the middle column of figure 1.1's flow chart.

1.1.1 Problem description

For the next generation of pressurised boiler components, a material has to be selected that can withstand yielding, creep and fatigue at temperatures up to 800 °C during long-term exposure while a good heat transfer over the component's wall is maintained (i.e. good heat conductivity and good oxidising resistance). For these conditions, high-alloyed steels do not suffice, so another metal-based material must be chosen. NEM, the sponsor of this project, has little experience with the use of other materials than steels and for that reason, a thorough evaluation of the properties and characteristics of this other metal-based material is needed. Also, this material must be a material approved by the design code, which is required for design approval by the design code's authorised inspector. It is the project sponsor's wish to use the material data directly into a proven design setup, so relations between material properties and service-induced parameters, like stress, temperature and time, are wanted. Another preference is the avoidance of extensive FE calculations for combinations of stress, temperature and time that will not lead to the desired lifetime of the boiler component or that will not meet the design code's limits.

1.1.2 Approach to solving the problem

First, literature research is performed to gather abundant, reproducible and explainable data. From these data, equations are formulated for the material properties, which depend on temperature, stress and/or time. The material properties and the validity of these equations are evaluated and discussed. Also, conservative factors enforced upon by the design code are applied to these equations to ensure a safe and reliable design.

A material model is built in FE software program ANSYS with the use of the equations. With the material model in ANSYS, the material can be implemented directly into the design. A Toolbox is constructed as well, that performs the same calculations as the material model in ANSYS, but additionally provides a fast overview of all material and design code criteria and considers the design code's conservative factors and limits, making extensive ANSYS calculations redundant.

1.2 Heat Recovery Steam Generator (HRSG)

The HRSG uses the heat of the gas turbine's exhaust gas to evaporate water. The steam is led through a steam generator to generate electrical power. The combined cycle process of gas turbine and steam turbine is especially used for power generation and is usually built for energy companies. The performance efficiency of the combined cycle can be up to 50 - 60 percent, which is equivalent to 20 to 350 MW depending on the size of the installation. The HSRG design can be for horizontal (see figure 1.2) or vertical gas flow and can be based on drum-type or 'once through' lay-out [1].



Figure 1.2: Horizontal HRSG, natural circulation. Exhaust gas flow from right to left. By courtesy of NEM Energy BV.

Regardless the design, the heat exchanger tube bundles of the HRSG are positioned perpendicular to the exhaust gas flow. The water/steam mixture is transported through the exchanger tubes by natural or forced circulation. The heat exchanger tubes are often finned tubes that are pre-fabricated in 'harps', which consists of two headers (i.e. kind of mini drums/vessels) that are connected by the exchanger tubes (see figure 1.3, left side), or consists of one header with returning tubes (see figure 1.3, right side). These harps can be combined in a pre-fabricated box, called a module. The harps or modules are installed inside the HRSG at site [1].

1.2.1 Next generation of materials for HRSG harps

At the moment, steels with higher chromium contents, like grade 91 (9Cr-1Mo-0.25V) are used in the first module of most HRSGs (i.e. closest to the inlet for the exhaust gasses), which is exposed to the highest temperatures. These higher chromium steels are proven to be reliable for at least 15 years of service at maximums of approximately 560 °C and 200 bar (20 MPa). Due to developments in the gas turbine industry, the temperature of the exhaust gasses has increased to circa 620 °C. Material AISI 304H is applied as material for the first module in the latest HRSGs to meet the conditions caused by this rise in temperature, though the long-term performance of this material is not yet certain. Other reasons of concern are the rather large coefficient of thermal expansion of austenitic stainless steel, that results in an increased degree of self-constraint, and the naturally formed oxide layer, that reduces the heat transfer.



Figure 1.3: Left: header with connection tubes on which the heat exchanger tubes will be installed; Right: return bends at the end of the heat exchanger tubes. By courtesy of NEM Energy BV.

1.3 Solar Boilers

Nowadays, it is preferred to generate power from renewable resources. Therefore, the concept of combining solar energy with steam power into a solar boiler has been developed as a viable solution.



Figure 1.4: 3D drawing of the solar receiver. By courtesy of NEM Energy BV.

The basic design is to transfer the solar radiation energy to molten salt inside heat exchanger tubes. The molten salt system in turn exchanges heat with water tubes to generate steam which is led to a steam generator to produce electrical power. A drawing of the receiver is given in figure 1.4 and the schematic representation of the complete installation is shown in figure 1.5.



Figure 1.5: Schematic representation of a solar boiler. By courtesy by NEM.

1.4 Material selection

The application of a material subject to cyclic loads at high temperatures requires the combination of material properties that ensures sufficient strength, ductility, creep-resistance, resistance against thermal cycling, and oxidation resistance. It is shown that combined solid-solution and precipitate-strengthened nickel-based alloys (NA), also known as superalloys, are providing the best combination of these material properties. By adjusting the chemical composition of the NAs and by applying sophisticated heat treatments, the properties of the material can be fine-tuned. The effects of the alloying elements in NAs are categorised in table 1A.

Taking the alloying effects into consideration and choosing a material that has been researched abundantly (i.e. quite some material data available and accepted by the design codes), the prime candidate for pressurised components is Alloy 617. Chapter 2 describes Alloy 617 and its properties more detailed. Perhaps more promising candidates for the pressurised components exist in regard to better performance or cost reduction, but these candidates are not examined that thoroughly yet to be adapted into the design code.

Alloying elements	Main feature for aqueous corrosion	Main feature for high-temperature applications	Other benefits	
Ni	Provides corrosion resistance to caustic solutions and dilute deaerated solutions of nonoxidizing mineral acids. Improves chloride SCC	Stabilization of austenitic phase. Provides precipitation of γ^\prime	Thermal stability and fabricability	
Cr	Provides resistance to oxidizing media Enhances localized corrosion resistance	Provides solid solution hardening Provides precipitation of M ₂₃ C ₆ , as benefit for notched rupture resistance		
Mo	Provides resistance to reducing media	Provides solid solution hardening; provides precipitation of M ₈ C		
	Enhances localized corrosion resistance			
W	Behaves similar to Mo but less effective	Provides solid solution hardening	Detrimental to thermal stability	
		Provides precipitation of M ₆ C		
AI		Suppress precipitation of η phase (Ni ₃ Ti)	Deoxidizer in melting process	
		Provides oxidation resistance		
Ti		Provides precipitation of γ'		
Nb. Ta		Provides precipitation of γ' and γ''		
Si		Provides oxidation resistance	Increases fluidity in casting process	
C	Affects detrimental effect for sensitization	Provides solid solution hardening	Mechanical properties	
		Provides precipitation of M ₂₃ C ₆ , M ₆ C, MC, etc., much precipitation of MC decreases precipitation of γ' and γ''		
Ν		Austenitic stabilizer	Thermal stability and mechanical properties	
Cu	Improves resistance to seawater			
B, Zr		Increases creep rupture strength		
<1.5.1.1.1.1.1.1.1.1.1.1.1.1.1.1.1.1.1.1		Suppress precipitation of n phase		
La, Ce		Provides oxidation resistance	Deoxidizer in melting process	

Table 1A: Effects of alloying elements on various properties of NAs [2].

1.5 ASME Design Code and choice for ASME

Design codes specify requirements and limits for design, fabrication, inspection and testing. Often these design codes are bond to the country where the installation is placed. For steam power installations, the applicable design codes are summarised in table 1B.

Country	Code	Issuing authority	Acceptance country
US	ASME Boiler and pressure Vessel Code	ASME	Canada, Emirate, France
	API 579	API	
	BS-1515, BS-5500, BS-7910	British Standard Institute	
UK	R5, R6	Britsh Energy	
Germany	FBH, KTA, TRD	Arbeitsgemeinschaft Druckbehalter	
Italy	ANCC		
Ne therl an ds	Regels Voor Toestellen	Dienst voor het Stoomvezen	
Sweden	Tryckkarls kimmissionen		
Australia	AS 1200:SAA Boiler Code		
Belgium	IBN Construction Code for Pressure Vessel		
Japan	MITI Code, JSME code	MITI and JSME	
France	RCC-M,RCC-MR	AFCEN	
India	IS-2825		
Europe	PED, EN 13445, EN 12952	European commission and AFNOR	1

Table 1B: Design codes per country.

More work is undertaken to harmonise the design codes in order to standardise the regulations, requirements and limits. At the moment, it is important to choose the design code by taking into account the application, the needs and the country's legislation.

The American Society of Mechanical Engineers (ASME) is the first organisation to develop a design code for pressurised components, and they started with the steam boilers, like HRSGs, in the ASME I. Many other design codes have found their basis in the ASME codes, although changes are made to these design codes to meet product specific or country enforced requirements.

The sponsor of this research is an ASME certified company having much experience in the design and manufacturing of HRSGs in accordance with the requirements and limits of the ASME design codes. Also, the ASME code is the most frequent used design code for the design and fabrication of steam boilers throughout the world. For these reasons, the emphasis is set on using the ASME design code in this research.

1.6 Report Structure

This report consists of this introduction chapter, a chapter dealing with Alloy 617 properties (chapter 2), which includes high-temperature damage mechanisms (i.e. creep) and fatigue and their interaction. Also, the design code requirements and limits are dealt with in chapter 2. From the equations, correlations and data presented in chapter 2, a Toolbox has been constructed in Microsoft Excel, which is described in chapter 3. Observations based on the output of the Toolbox are shown in the later part of chapter 3, together with the Toolbox's limitations. The material model developed for the use in ANSYS and other data required for running calculations and simulations in ANSYS are handled in chapter 4. Comparisons between the results from the Toolbox and those from ANSYS are shown in chapter 5, which form the validation of the Toolbox. This research has left some aspects for discussion in chapter 6. Chapter 7 contains the conclusions from this research.

Appendices to this report have been added with background information on creep in appendix A, on fatigue in appendix B and about the procedure for fit-for-service (FFS) assessment in appendix C. Appendix D covers the explanations and extended calculations of some of the mathematic equations used in this report. The Bree diagram for the determination of ratchetting occurrence is explained in detail in appendix E and the requirements and limits enforced upon by the ASME design code are handled in appendix F. Appendix G describes a possible test setup for the testing of creep-fatigue interaction.

2. Alloy 617

Alloy 617 is a nickel-based (super)alloy (NA) with major alloying elements chromium, cobalt and molybdenum [3-5] which provide solid-solution strengthening. Parallel to that, precipitation strengthening is provided by intra- and inter-granular γ' -precipitates (Ni₃(Al,Ti)) and carbides. Alloy 617 has an exceptional combination of high-temperature strength and oxidation resistance [5,6], since nickel and chromium improve the alloy's resistance to oxidising and reducing media, while aluminium and chromium contribute to high temperature oxidation resistance [3,5,6]. Also, the creep-rupture strength is sufficient up to 980 °C [3]. The effect of alloying elements on the properties of NAs is shown in table 1A in paragraph 1.4.

This alloy has good creep strength at high temperatures, good weldability, and good cyclic oxidation and carburisation resistance [3]. In comparison to other material candidates, like austenitic stainless steels, Alloy 617 has a lower coefficient of thermal expansion and a higher thermal conductivity. Other benefits of this material are that it retains its toughness after long-term high-temperature service and that intermetallic phases can be avoided which might cause embrittlement [4].

One of the challenges for Alloy 617 is (surface) oxidation though the resistance is quite reasonable. This alloy is more prone to grain boundary oxidation due to the formation of aluminium-rich grain boundary oxides. This kind of oxidation is especially disadvantageous in heat exchanger designs since it may affect heat transfer and cause initiation sites for possible cracks.

2.1 Properties of Alloy 617

2.1.1 Chemical composition alterations

To ensure the formation of the desired microstructure and minimise deviations in the material's performance, efforts have been taken to limit the chemical composition range as set in the material standards commonly used [7]. A controlled material specification (CMS) of Alloy 617 is suggested with the aim to reduce the data scatter and to achieve and maintain the high temperature beneficial material properties [8]. This CMS is based on computational modelling.

Computational thermodynamic modelling predicts the possible formation of the unwanted sigma (σ) phase ((Ni,Co)_x(Cr,Mo)_y with x,y = 1...7) for different chemistries of Alloy 617 [3,8]. The maximum chemistry (i.e. using the maximum contents of the alloying elements) of the used material standard for Alloy 617 [7] is shown in the left figure of figure 2.1. The σ -phase can cause a detrimental decrease of toughness at temperatures below 1050 °C. The alloying elements chromium, cobalt, molybdenum, and matrix element nickel promote σ -phase formation and their concentrations in the σ -phase vary with temperature.

The minimum amount of alloying elements in the chemistry of the used material standard [7] cannot be used, since the desired γ' -precipitates are not present and the amount of carbides (required for grain refining and dislocation barriers) is limited (see right figure of figure 2.1).



Figure 2.1: Modelling result of second phases at various temperatures in equilibrium for the maximum contents (left figure) and minimum contents (right figure) of alloying elements in the composition of Alloy 617 according to the material standard [8].

The challenge in choosing an optimal chemistry is to avoid σ -phase, while maintaining the contents of the strengthening elements as high as possible [8]. An example of a phase diagram evolved from the computational modelling is shown in figure 2.2. In this figure the weight percentages of the alloying elements are 15.0% Co, 8.0% Mo, 1.40% Al, 0.60% Ti, 0.10% C, and 0.04% N. The chromium content has been varied between the minimum and the maximum values as specified by the material standards. The outcome of modelling with these alloy element contents shows that σ -phase does not appear if the chromium content is maximised at 22.7 wt% [8].



Figure 2.2: Calculated equilibrium phase for possible CMS Alloy 617 chemistry combinations by adjusting the Crcontent [8].

Various chemical compositions have been calculated based on these modelled phase diagrams while trying to maximise strength and to avoid/limit σ -phase formation. From this data, the controlled material specification (CMS) is determined as shown in table 2A. Although the chance of σ -phase formation is present, the higher chromium level is required to ensure sufficient creep resistance and oxidising resistance. Heat treatment can be performed to avoid the formation of σ -phase [8].

Heat	Ni	Cr	Co	Mo	Fe	Mn	Al	C
ASTM Max	-	24.0	15.0	10.0	3.0	1.0	1.5	0.15
ASTM Min	44.5	20.0	10.0	8.0		1 1 1	0.8	0.05
CMS Max		24.0	15.0	10.0	1.0	1.0	1.40	0.10
CMS Min	44.5	22.0	13.0	9.0	-	-	1.20	0.07
Heat	Cu	Si	S	Ti	P	B	N	Nb
ASTM Max	0.5	1.0	0.015	0.6		0.006		-
ASTM Min	-	-	-	-	-	-	: -	-
CMS Max	0.2	0.3	0.008	0.60	0.010	0.005	0.040	
CMS Min	-	-	-	0.40	-	0.002	: -	

Table 2A: ASTM/ASME standard chemistry and initial CMS for Alloy 617 (wt.%) [8].

Some concerns exist for the CMS chemistry of Alloy 617 [8]:

- The chemistry range of iron should be between 1.5 2.0%, since a maximum of 1.0% is too stringent for economical production.
- The ranges for aluminium (1.20 1.40%) and titanium (0.40 0.60%) are that narrow that a cost increase may be expected.
- High levels of chromium, cobalt, and molybdenum are required to improve creep strength, but at the same time these alloying elements make processing of the material (i.e. hotworking) more challenging.

2.1.2 Grain size influence

The grain size is important, since grain size no. 6 according to ASTM E112 (approximately 45 μ m) or coarser are preferred for better creep resistance, because diffusion of vacancies and atoms is faster via grain boundaries than through the grain, but finer grains are wanted for better fatigue resistance because the grain boundaries act as fatigue damage blockers. An optimal grain size is desired to cope with both creep and fatigue mechanisms [4,9]. The carbides in Alloy 617 are grain size controllers [6].

Creep tests have been performed on samples with an average grain size of circa 180 µm (ASTM E112, no. 2 grain size), of approximately 95 μ m (ASTM no. 3.5) and of around 20 μ m (ASTM no. 8) [9]. The main observation is that ASTM no. 3.5 samples gave a stress exponent of approximately 5-6 regardless the stress magnitude (5 – 100 MPa) and at temperatures of 850 – 1050 °C. This suggests that power law dislocation climb-glide creep is the dominant creep mechanism. The activation energy depends on the (amount of) precipitates in the material, though it is found to be higher than that of nickel selfdiffusion [9]. ASTM no. 8 (finer grains) sample testing resulted in a stress exponent of around 3 for stress magnitudes of 5 – 100 MPa and at temperatures of 850 – 1050 °C. This is most likely caused by a combination of dislocation and diffusional creep [9]. This combination of creep mechanisms can better be avoided since these occur in parallel and their combined action makes the creep prediction more complex. ASTM no. 2 samples (coarse grains) show a stress exponent of circa 3 for the lower stresses at temperatures between 850 – 950 °C probably caused by a solute drag creep mechanism. With increasing stress, this stress exponent shifts from 3 towards 5 probably because power-law creep becomes more present. For higher temperatures, the stress exponent shifts towards a value of 5 likely due to power-law dislocation climb-glide. The coarser ASTM no. 2 grains would be more favourable for creep characteristics, but the dominant creep mechanism is also a combination of two mechanisms for which the ratio between them is dependent on stress and temperature. Also, the coarser grains reduce fatigue resistance.

The optimal grain size for Alloy 617 in creep-fatigue conditions is ASTM no. 5 or 6 (45.0 μ m – 63.5 μ m). These grains are sufficiently coarse to obtain good creep resistance and to avoid complex interactions between creep mechanisms while remaining small enough to ensure good fatigue properties.

2.2 Plastic behaviour

Mechanical strains, either caused by external loads or (thermal) constraints may lead to plastic yielding. The stress at which plastic strain starts to develop depends on the temperature. Therefore, knowledge of the proportionality limit, the 0.2% yield strength and the ultimate tensile strength related to the temperature is important for design purposes.

2.2.1 Proportionality Limit

The proportionality limit (*P*) is the highest stress at which Hooke's Law (i.e. stress is proportional to strain) is still valid. The slope of the linear part of the stress-strain graph is equal to the elastic modulus. Below *P* the material is considered to deform purely elastically [10]. Above *P* the material starts to deform plastically. Because the proportionality limit is hard to determine, usually the yield strength at 0.2% offset (σ_y) is used, since that can be determined much easier (e.g. by using less accurate test equipment). Disadvantage of the σ_y is that there is already 0.2% plastic strain. See for a schematic representation figure 2.3.



Figure 2.3: Proportional limit (P), σ_v and elastic and plastic deformation [10].

The *P* and the σ_y together with the Young's modulus (E) and the ultimate tensile strength (UTS) of Alloy 617 have been investigated in the temperature range 800 °C – 950 °C from conventional tensile tests [11] and the results are shown in table 2B.

Temp. (ºC)	E (MPa)	Yield Stress, σ _y (MPa)	Proportional Elastic Limit, σ _p (MPa)	UTS (MPa)
800	157000	259.2	242.0	327.0
85 <mark>0</mark>	153000	212.8	181.0	235.5
900	149000	168.9	139.3	190.7
950	144000	132.3	107.3	148.7

Table 2B: strength properties of Alloy 617 [11].

The ratios between P and σ_y are 0.93 for 800 °C, 0.85 for 850 °C, 0.83 for 900 °C, and 0.81 for 950 °C. Other data for P in Alloy 617 are not available, but extrapolation of these data leads to the observation that the P in Alloy 617 up to 750 °C is certainly minimal 0.95 σ_y . The importance of this information lays in the fact that design codes prohibit the use of the material above a certain allowable stress, which is a ratio/percentage of the yield strength (e.g. $\frac{2}{3}$ or 90%) at a specific temperature as one of the criteria. For that reason, plastic deformation caused by a primary stress (i.e. directly applied stress) is not expected to occur for Alloy 617 at temperatures up to 800 °C. Above 800 °C, the maximum allowable stress is limited due to time-dependent (i.e. creep) mechanisms, which results in a significantly lower stress than the proportionality limit.

2.2.2 Yield Strength

The temperature dependency of the yield strength has been determined from the data points from several heats [12]. The data has been normalised by dividing the strength data of each heat by the average strength value at room temperature of that heat.

First, a best fit trend curve ($R_Y(T)$) is constructed from the normalised data. Then, the temperaturedependent yield strength $S_Y(T)$ is defined as the multiplication of $R_Y(T)$ and the minimum yield strength at room temperature (S_Y) (i.e. 240 MPa for Alloy 617 according to the material standard used [7]). The best fit is obtained from a piecewise continuous exponential decay function [13,14]:

$R_Y(T) = $	$\int a_1 + b_1 \exp[c_1 T]; T \le T_0$	(2.24)
	$\{a_2 + b_2 \exp[c_2 T]; T > T_0$	(2.2A)

in which T is the temperature (in °C), and a_1 , a_2 , b_1 , b_2 , c_1 , c_2 , and T_0 are determined from the data. By multiplying the best fit curve equation with the minimum yield strength at room temperature, a conservative lower bound is obtained as shown by the dashed green line in figure 2.4.

Since all data points are significantly above the lower-bound trend line, the parameters may be adjusted to obtain a more representative (and less conservative) lower bound (see figure 2.4). First, from the best fit curve from the data (represented by the solid black curve in figure 2.4) the 95% prediction bound is determined (dashed blue line) [12]. Then the minimum yield strength values at several temperatures as given by the stress tables from ASME II, part D, (shown as red asterisks) are extrapolated by using the equation as given in equation (2.2A) [12] (orange line in figure 2.4).



Figure 2.4: Data points for the yield strength obtained from tests (green asterisks), trendline (black line) and the 95% prediction bound (dashed blue line) and the values for the minimum yield strength from ASME (red asterisks and yellow line) [12].

The best fit parameters for the yield strength using the ASME II method curve and the 95% lower prediction bound of the data are shown in table 2C. The values from the ASME method still have to be multiplied with the S_Y at room temperature (i.e. 240 MPa). The 95% lower prediction bound is determined by lowering the best-fit curve from the data by 7.764E+01 MPa. The 95% lower prediction bound (blue dashed curve) intersects the ASME II method curve (orange solid curve) at about 825 °C.

Parameter	ASME method	Data
<i>a</i> ₁	6.846E-01	1.605E+02
b_1	3.610E-01	1.131E+02
<i>C</i> ₁	-5.752E-03	-3.898E-03
T_0	8.275E+02	8.246E+02
<i>a</i> ₂	-2.886E-01	-2.111E+02
<i>b</i> ₂	9.983E+00	2.872E+03
<i>C</i> ₂	-2.810E-03	-2.465E-03

Table 2C: Best-fit parameter coefficients for the yield strength [12].

Evaluation of the graph shows that up to 825 °C, the ASME II method (i.e. orange solid curve) is actually too conservative and the 95% prediction lower-bound curve from the data is still sufficiently conservative and represents the actual data more accurately. Above 825 °C, the ASME II method gives a better illustration of the minimum obtained yield strengths from the tests though remaining

conservative, while the 95% prediction lower bound underestimates the yield strength significantly. Therefore, it suffices to express the minimum yield strength at temperatures below 825 °C with the 95% prediction lower-bound best-fit curve (blue dashed curve) and above 825 °C with the ASME II method (i.e. orange solid curve). The result is shown in figure 2.5.



Figure 2.5: Minimum yield strength as function of temperature.

A statistical evaluation of this lower-bound minimum yield strength plot as shown in figure 2.5 has not been performed, since the design code's conservative factors are applied on this plot as is described later in this report.

2.2.3 Tensile Strength

The tensile strength is determined in a similar manner as the yield strength. After normalisation of the strength data (i.e. dividing the strength data of each heat by the average strength value at room temperature of that heat), a best-fit curve ($R_T(T)$) is determined [14]:

$$R_T(T) = \begin{cases} a_3 + b_3 T; T \le T_0 \\ a_4 + b_4 \exp[c_4 T]; T > T_0 \end{cases}$$
(2.2B)

in which T is the temperature (in °C), and a_3 , a_4 , b_3 , b_4 , c_4 , and T_0 are determined from the data. The average tensile strength at temperature $S_T(T)$ is defined as $S_T(T) = S_T R_T(T)$ [12], in which S_T is the minimum tensile strength at room temperature (i.e. 655 MPa according to ASME II).

Plotting of the best-fit curves through the tensile stress data points is illustrated in figure 2.6. The best fit curve for the measured data is indicated by the solid black curve [12]. From this measured data, the 95% prediction lower bound (in figure 2.6 indicated as the 'constant offset' shown by the dashed green line) is determined by lowering the stress by approximately 43 MPa. This lower-bound curve coincides fairly well with the curve determined by the ASME method (solid orange curve). At lower temperatures, the 95% prediction lower bound (dashed green line) is less conservative than the orange ASME method curve. At higher temperatures, this is vice versa.

The sets of parameters of the 95% prediction lower bound (dashed green line) and the ASME method (solid orange line) are obtained and these are shown in table 2D [12]. The final calculated value of the ASME method has to be multiplied with the tensile strength at room temperature (i.e. 655 MPa) [7].



Figure 2.6: Adjusted best-fit curve for tensile strength [12].

Parameter	ASME method	Data
<i>a</i> ₃	1.008E+00	7.287E+02
<i>b</i> ₃	-3.989E-04	-2.935E-01
T ₀	7.241E+02	7.227E+02
<i>a</i> ₄	-3.916E-02	-7.147E+01
b_4	2.185E+01	1.746E+04
C ₄	-4.641E-03	-4.691E-03

Table 2D: Best-fit parameter coefficients for the tensile strength [12].

For the yield strength in the previous paragraph, the ASME method is actually too conservative and the least conservative mixture of ASME and 95% prediction lower bound is chosen. For the tensile strength, this observation cannot be made and the most conservative mixture is chosen for certainty.

As can be observed in figure 2.6 and considering the most conservative situation, the ASME II method (solid orange curve) gives lower values for the minimum tensile strength for temperatures below 723 °C, while the 95% prediction lower bound of the measured data (dashed green curve) provides lower values at temperatures above 723 °C [12]. For that reason, it is convenient to express the minimum yield strength at temperatures below 723 °C with the ASME II method and above 723 °C with the 95% prediction lower-bound best-fit curve. The result is shown in figure 2.7.

A statistical evaluation of the ultimate tensile strength lower-bound plot as shown in figure 2.7 is not done, since the design code's conservative factors are applied on this plot in a later stadium.



Figure 2.7: Minimum tensile strength as function of temperature.

2.2.4 Yield strength and tensile strength results combined

The best-fit curves and equations obtained for both yield strength and tensile strength versus temperature provide a representative and conservative method to calculate the minimum yield and tensile strength at any temperature between room temperature and 1,000 °C. In figure 2.8, both minimum yield strength and minimum tensile strength are shown as a function of temperature. The curves for the yield strength and tensile strength are quite similar to those shown in figure 2.5.



Figure 2.8: Minimum yield and tensile strength as function of temperature.

2.3 Creep behaviour

At high temperatures (i.e. homologous temperatures $T_h = \frac{T}{T_{melt}} > 0.5$), roughly four types of creep mechanisms can be distinguished. A deformation mechanism map for Alloy 617 is not available, though it is expected that it looks quite similar to the deformation mechanism map of MAR-M200, which is just as Alloy 617 a solid-solution and precipitate strengthened nickel-based superalloy. A deformation mechanism map of MAR-M200, is shown in figure 2.9. At high stresses (roughly $\frac{\tau}{\mu}$ > 10^{-2} , where $\frac{\tau}{\mu}$ is the normalised stress in which τ is the applied shear stress and μ is the shear modulus at temperature), dislocation glide is the dominant mechanism. Dislocation glide is more a plasticity mechanism than a creep mechanism and it involves the movement of dislocations along slip planes and the passing of obstacles (e.g. solute atoms, precipitates, other dislocations) by thermal activation. At intermediate stress levels (approximately $10^{-3} < \frac{\tau}{\mu} < 10^{-2}$), dislocation creep is the prevailing mechanism, which is a combination of dislocation glide and vacancy diffusion. This mechanism is commonly accompanied with a steady-state creep rate ($\dot{\epsilon}_c$) that is a balance between the strain hardening rate and the recovery softening rate. At low stresses (basically $\frac{\tau}{\mu} < 10^{-3}$), two creep mechanisms can occur, which are diffusion creep (i.e. flow of vacancies and interstitials through grains or along grain boundaries) and grain boundary sliding that involves a shear process along the grain boundaries. Increasing temperature and/or decreasing strain rate promotes grain boundary sliding, making it the dominant creep mechanism. At $T_H > 0.9$ and normalised stresses above 7 $\cdot 10^{-4}$, dynamic recrystallisation occurs, which is defined as the nucleation and growth of new grain grains induced by the deformation of the material and the temperature. In appendix A, more detailed information about these creep mechanisms is provided. The data used for the determination of the creep behaviour of Alloy 617 is for the greater part taken from data determined for Alloy 617 with average grain sizes of 45.0 μ m – 63.5 μ m (i.e. no. 5 and 6 according to ASTM E112).



Figure 2.9: Deformation mechanism map for MAR-M200 with grain-size of 100 μ m [15]. $\sigma_{s} = \tau$

2.3.1 Determination of creep mechanism

Assuming that the deformation mechanism map for Alloy 617 is quite similar to that of MAR-M200 in figure 2.9, three main mechanisms can be distinguished, i.e. plasticity, diffusional creep and power-law creep. Plasticity is commonly avoided, since design codes specify maximum allowable stresses for the design calculations. This is normally a ratio/percentage of the temperature-dependent yield strength, which is often $\frac{3}{3}$ or 90% depending on the material [16-18]. At the larger part of the temperature range for Alloy 617, the yield strength is 120 – 240 MPa, which leads to a maximum allowable stress of approximately 110 – 215 MPa. The shear modulus (μ) of Alloy 617 at higher temperatures is in the range of 60 – 80 GPa. The ratio between the maximum allowable stress (σ) and μ is circa $2 \cdot 10^{-3} - 3 \cdot 10^{-3}$, hence no plasticity. For the occurrence of creep, this ratio of $2 \cdot 10^{-3} - 3 \cdot 10^{-3}$ between σ and μ results in diffusional creep at temperatures up to circa 800 °C and in power-law creep at temperatures of approximately 800 °C and over (see also figure 2.9).

2.3.2 Creep rate equation

The Zener-Hollomon parameter (Z) can be used to normalise creep data, so temperature and strain rate can be related directly to each other [19-24]:

$$Z = \dot{\varepsilon}_c \exp\left(\frac{Q_c}{R \cdot T_K}\right) \quad , \tag{2.3A}$$

in which $\dot{\varepsilon}_c$ is the steady state creep strain rate, Q_c is the activation energy for creep, R is the gas constant, and T_K is the absolute temperature (in K).

The Zener-Holloman equation is related to Norton's creep law (also known as power-law creep):

$$Z = B_c \left(\frac{\sigma}{\mu}\right)^n \quad , \tag{2.3B}$$

in which B_c is a temperature-dependent constant, σ is the applied stress, μ is the shear modulus at temperature T_K , and n is a material exponent. B_c may be described as:

$$B_c = \frac{D \cdot \mu \cdot b}{k_B T_K} \qquad , \tag{2.3C}$$

in which D is the diffusion coefficient at temperature, b is the Burger's vector, k_B is the Boltzmann constant, T_K is the absolute temperature (in K).

The $\dot{\varepsilon}_c$ is then rewritten with a power-law dislocation part and a diffusional part as:

$$\dot{\varepsilon}_c = B_c \left(\frac{\sigma}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \qquad (2.3D)$$

 Q_c is the activation energy for creep which is actually the activation energy for self-diffusion that has been determined at approximately 410 kJ/mol for Alloy 617 [19,25-28], as can be seen in figure 2.10 for reference. The stress exponent n decreases from circa 13 at around 600 °C [29], to approximately 10 at circa 700 °C [21], and going to 8.3 at 750 °C, to 6.9 at 800 °C, to 5.9 at 900 °C and 5.1 at 1,000 °C [19] (see also figure 2.11).



Figure 2.10: Creep activation energy (Q_c) determination for Alloy 617 from the slope of creep strain rate vs. $1/T_K$ [25].



Figure 2.11: Relation between creep strength and creep strain rate of Alloy 617 [19].

An example calculation is described in paragraph D1 of Appendix D.

2.3.2.2 Temperature dependency of creep equation parameters

The shear modulus (μ) at several temperatures is determined [28] and plotted (see figure 2.12), and the corresponding equation is (in GPa):

$$\mu(T) = -9.5044 \cdot 10^{-6}T^2 - 1.8855 \cdot 10^{-2}T + 8.1501 \cdot 10^1 \quad , \tag{2.3E}$$

with the temperature T in degrees Celsius.



Figure 2.12: Shear modulus as a function of temperature.

The stress exponent n is approximately 10 for Alloy 617 at approximately $T_h = 0.6$, which is between 690 – 720 °C. For the determination of an equation for n, the average value of T_h of 0.6 has been taken (i.e. 705 °C) and combined with other n values found for higher temperatures [15]. The resulting equation for the temperature range 706 – 801 °C is:

$$n(T) = 1.3707 \cdot 10^{-4}T^2 - 2.3924 \cdot 10^{-1}T + 1.1058 \cdot 10^2 \quad , \qquad (2.3F1)$$

with T in degrees Celsius. In the temperature range above 801 °C, the equation for n is:

$$n(T) = 1.0000 \cdot 10^{-5} T^2 - 2.7000 \cdot 10^{-2} T + 2.2100 \cdot 10^1 \quad , \quad (2.3F2)$$

with T in degrees Celsius and these equations are shown in figure 2.13.



Figure 2.13: Stress exponent as a function of temperature.



The data [15] is also used to determine an equation for material parameter B_c (see figure 2.14):

Figure 2.14: Material parameter B_c (*in* $[h^{-1}]$) *as a function of the temperature.*

2.3.3 Influence of γ' particles on creep rate

Creep behaviour in single phase alloys (like pure nickel) behaves in a power-law relationship, for which the minimum creep rate ($\dot{\varepsilon}_{c.min}$) can be described as [19,30-33]:

$$\dot{\varepsilon}_{c,min} = B_c \left(\frac{\sigma}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \quad , \tag{2.3H}$$

in which the parameters are the same as those in equation (2.3D).

The *n* values and the activation energy for self-diffusion (Q_c) for pure nickel are smaller than those for precipitation-hardened alloys as Alloy 617, which are aged to contain γ' -precipitates [34,35]. A threshold stress (σ_0) is introduced to the power-law part of the minimum creep strain rate equation to account for the influence of precipitates on the movement of mobile dislocations [33,36]:

$$\dot{\varepsilon}_{c,min} = B_c \left(\frac{\sigma - \sigma_0}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right)$$
(2.31)

This threshold behaviour is common in γ - γ' superalloys [37] when Orowan bowing is active (i.e. bowing of a dislocation line around a particle until the critical curvature is reached which causes a dislocation loop around the particle) besides the thermally activated creep mechanisms dislocation gliding (i.e. dislocations that move along slip planes overcome barriers by thermal activation), dislocation climbing (i.e. edge dislocations avoid barriers by moving to another lattice plane) or shearing (i.e. dislocations cut/shear through the precipitate and move onwards) [38]. At 750 °C and below, Orowan bowing is the main dislocation hindering process, which is shown by the present threshold stress (i.e. it is difficult for dislocations to bow around the precipitates). At 800 °C and above, the threshold stress is absent, which indicates that Orowan bowing occurs quite easy and the thermally activated creep mechanisms are dominant or that the precipitates have disappeared (i.e. dissolved).

There are two methods to determine σ_0 . The first method is using creep data to compare Alloy 617 behaviour at different temperatures. As shown in figure 2.15, the Zener-Hollomon parameter as a function of the normalised stress data at several temperatures is expected to overlap, so one plot can be constructed. This is done by shifting the data determined at 750 °C as shown in the left graph of figure 2.15 to the left by circa 65 MPa (figure 2.15, right figure). This data shift is σ_0 . The drawback of this method is that this analysis is difficult since the microstructure is evolving during creep, though it has been observed that for Alloy 617 the σ_0 remains reasonably constant over time [19].



Figure 2.15: Left: unaltered test data; Right: test data with σ_0 for 750 °C test data [39].

The other method is the stress-drop test in which a conventional lever arm creep frame configuration is used [19,34,35,40]. The threshold stress is determined by evaluating strain transients that result from small stress reductions. First, an initial stress (σ_i) is applied to induce a constant forward creep rate (graph up to point A in figure 2.16). The stress drop causes an instantaneous strain reduction (A to B), followed by relaxation (i.e. a time dependent contraction; B to C). After C, there is a period Δt of constant strain, which the mechanical result strain only (i.e. neither forward nor backward creep) until forward creep is re-established due to the applied constant stress (at D) [19].



Figure 2.16: Strain evolution during a stress-drop test caused by the stress reduction [19].

Subsequent stress reductions are applied until the strain transients following a stress reduction are not distinguishable anymore from the re-established strain (i.e. no strain increase observed anymore after a certain amount of time). The stress at which this occurs is the onset stress for creep ($\sigma_{c,min}$) caused by Orowan bowing:

$$\sigma_{c,min} = \sigma_i - \sum \Delta \sigma \qquad (2.3J)$$

Plotting $\sigma_{c,min}$ versus time segment (Δt), which is the time interval between stress drop and the onset of new creep accumulation (see figure 2.16), allows for a quantitative determination of σ_0 , since at the asymptote of $\Delta t \rightarrow \infty$, the $\sigma_{c,min} \rightarrow \sigma_0$ The extrapolation of σ_0 from the asymptote may lead to a significant error due to strain resolution limitations. For that reason, a stress-decrement model for recovery controlled creep is used to fit the $\sigma_{c,min}$ versus $\sum \Delta t$ curve [41], i.e.:

$$\sum \Delta t = \frac{\mu^2 b^2}{c_K} \left[\left(\sigma_{c,min} - \sigma_0 \right)^{-2} - (\sigma_i - \sigma_0)^{-2} \right] , \qquad (2.3K)$$

in which C_K (in meter) is a kinetic constant, b is the Burger's vector and the other parameters are described previously in this paragraph.

2.3.3.1 Stress influence

Creep in Alloy 617 requires the mobile dislocations to bypass the γ' -precipitates by Orowan bowing or to continue by thermally activated creep mechanisms, like climb. The stresses required to activate precipitate shearing are that high, that this creep mechanism is not expected [42]. This is confirmed by tests, which show the absence of sheared precipitates [19]. Creep is retarded by the precipitates which results in an improved creep strength. This retarding may be caused by the precipitate– dislocation interaction that attracts and detaches dislocations to/from precipitates [43,44], i.e. Orowan bowing. Depending on this interaction, the creep mechanism is detachment-controlled or climb-controlled. This influence is indicated by the 'stress relaxation parameter' κ , which can vary between zero (full attraction) and unity (attraction-free). For a small volume fraction of precipitates ($f_{\nu} < 0.13$), the normalised threshold stress, which is the ratio between the shear stress (τ) and Orowan stress (i.e. the combined shear modulus (G), Burgers vector (b) and precipitate spacing (2λ)), has been plotted as a function of the relaxation parameter (κ) in figure 2.17 for both detachment-controlled and climb-controlled mechanism.



Figure 2.17: Normalised threshold stress for detachment controlled creep and for climb controlled creep as function of the stress relaxation parameter [44]. $G = \mu$ and $2\lambda = L_S$.

The detachment-controlled stress depends on the shear stress required for Orowan bowing (τ_{Or}):

$$\tau_{Or} = \frac{\mu b}{L_S} \qquad , \tag{2.3L}$$

in which μ is the shear modulus, b is the Burgers vector (i.e. 254 pm for Alloy 617) and L_S is the edgeto-edge spacing between the precipitates. The shear stress can be converted to a normal stress by multiplication with the Taylor factor C_T (i.e. the relation between uniaxial yield strength and resolved shear stress in a poly-crystalline material assuming that all the grains deform uniformly). C_T is 3.06 for FCC alloys [19,43,45,46] and is needed to determine the stress required to activate Orowan bowing of dislocations around the γ' -precipitates. The edge-to-edge spacing (L_S) depends on the fraction volume of precipitates (f_v) and the precipitate diameter (d_p) [42,47]:

$$L_S = d_p \sqrt{\frac{8}{3\pi f_v}} - d_p \qquad (2.3M)$$

Thus, for a fixed volume fraction of precipitates, a smaller precipitate diameter results in a smaller edge-to-edge spacing L_S . The smaller L_S in turn leads to a higher τ_{Or} [19]. This results into a larger threshold stress and thus to a better creep resistance. When less precipitates are present (i.e. larger precipitates for a fixed volume fraction of precipitates), less precipitate-dislocation interaction exists and the stress relaxation parameter κ diminishes, in turn lowering the threshold stress. The shear stress (τ) to overcome the threshold stress becomes then [44]:

$$\tau = \sqrt{1 - k^2} \cdot \tau_{Or} \tag{2.3N}$$

For this reason, many small precipitates are preferred over less larger ones for a fixed volume fraction of precipitates.

Diffusion-controlled dislocation climb without precipitate interaction (i.e. he relaxation parameter κ is unity (=1)) can be equated as:

$$\tau_{climb} = 0.4\tau_{Or} \qquad , \qquad (2.30)$$

in which τ_{climb} is the shear stress required to increase the dislocation line length and climb around a precipitate and τ_{Or} is the Orowan stress. In the presence of precipitates, κ is less than unity and τ_{climb} becomes [48,49]:

$$\tau_{climb} = 0.4(\kappa)^{5/2} \cdot \tau_{Or} \qquad , \qquad (2.3P)$$

This trend is shown in figure 2.17. For this volume fraction of precipitates, climb limits creep more than Orowan bowing only when the stress relaxation parameter is going towards unity. In Alloy 617, many fine and wide-spread γ' - precipitates are present, resulting in a stress relaxation parameter in the proximity of zero. For that reason, as can be seen in figure 2.17, the threshold stress for the Orowan bowing creep mechanism is relatively high, while that of climb is almost non-existent. Therefore, creep is dominated by Orowan bowing in Alloy 617.

2.3.3.2 Consideration of Orowan threshold stress for determination of creep.

The Orowan bowing threshold stress has been incorporated in the literature data already for the determination of the creep rate as a function of the stress. For that reason, the stress used in the creep rate equation (2.3D) does not require any adjustment and may be used directly.

2.3.4 Stress relaxation

The stress relaxation process is a creep process where part of the elastic strain transforms into inelastic strain due to creep. The remaining elastic strain causes a resulting stress (i.e. Hooke's law applies), which reduces in time with the creep strain developing and elastic strain diminishing. This reduction in stress magnitude is known as stress relaxation. The stress relaxation rate may be calculated with a similar equation as for the creep strain rate expressed in equation (2.3D), though the determination is more challenging, since the stress is declining over time. The relation between the diminishing creep strain rate and diminishing stress can be described in a numerical manner and in an analytical manner. The numerical method is solving equation (2.3D) for subsequent time-steps that are sufficiently small to guarantee an adequate accuracy and calculating the creep strain accumulated until that time-step. The analytical method is an integration equation of equation (2.3D) in which the creep strain ($\varepsilon_{c,s}$) is calculated as a function of time (t):

$$\varepsilon_{c,s} = \varepsilon_0 - \left((1-n) \left(\frac{\varepsilon_0^{(1-n)}}{1-n} - B_c \cdot \left(\frac{E}{\mu} \right)^n \cdot \exp\left(\frac{-Q_c}{R \cdot T_K} \right) \cdot t \right) \right)^{\frac{1}{1-n}},$$
(2.3Q)

in which most terms are the same as in equation (2.3D) and additional are the elastic modulus (E) and the strain at start of the relaxation process (ε_0). The mathematics for the construction of equation (2.3Q) are shown in paragraph D2 of appendix D. The numerical method is used in the remainder of this report because most FE software programs and the design code also use the numerical method [16], which is considered as easier to perform and sufficiently accurate.

2.4 Time to rupture behaviour

In figure 2.18, the Larson-Miller Parameter (*LMP*) is shown as a function of stress. This plot has been developed from the results of uniaxial tests. The *LMP* depends on time to rupture (t_R ; in figure 2.18 indicated as 't') and absolute temperature (T in figure 2.18, given in Kelvin).



Figure 2.18: Larson-Miller plot for Alloy 617 [19].
The average creep rupture strength ($\sigma_{R,avg}$) is adequately approximated by [19]:

$$\sigma_{R,avg} = 4.5314 \cdot 10^5 \exp(-3.41 \cdot 10^{-4} LMP) \quad , \tag{2.4A}$$

with the Larson-Miller Parameters (LMP) equated as:

$$LMP = T_K(\log t_R + 20) \qquad , \qquad (2.4B)$$

in which T_K is the absolute temperature in Kelvin and t_R is the time to rupture in hours.

For designing purposes, it is recommended to use minimum creep rupture strength values instead of average creep rupture strength values. Using the data plots shown in figure 2.18 [19], figure 2.19 [50] and figure 2.20 [29], the average data plots and the data points below these plots are compared at several times to rupture and the deviations between them are determined. In table 2E, the deviations are given of the stress to rupture value of the lowest situated data points in comparison to the average value. A percentage is chosen instead of an absolute value in order to deal with the figure's log scale. The deviation percentages at 595 °C and 925 °C from figure 2.20 are not regarded for the evaluation, since solely data from bar material are available.



Figure 2.19: Stress versus 1% creep strain and time to rupture curves [50].

Table 2E: Maximum deviations of the lower data points in comparison to the average value from figures 2.22, 2.23 and 2.24.

Temp.	Deviation
[°C]	[%]
650	20
760	20
850-870	20
980-1000	20



Figure 2.20: Rupture strength of Inconel 617 [29].

A maximum deviation of 20% has been observed for creep rupture strength values below the average values for each temperature between 650 °C and 1,000 °C. Considering this 20% deviation in equation (2.4A), the minimum creep rupture strength $\sigma_{R,min}$ may be equated as:

 $\sigma_{R,min} = 0.8 \cdot \sigma_{R,avg} = 3.6251 \cdot 10^5 \exp(-3.41 \cdot 10^{-4} LMP) \quad . \tag{2.4C}$

A statistical evaluation of the data of the minimum creep rupture strength is not done, since the design code's conservative factors are applied at a later stage.

2.5 Fatigue

Cyclic loads due to transients in temperature and/or pressure are common in many pressurised components of HRSGs during service and start-stops. For that reason, fatigue can be a dominant damage mechanism instead of creep or together with creep. Since the maximum allowable stress by any design code has to be lower than the yield strength of a material determined at the highest temperature, the deformations due to these temperature-induced and/or pressure-induced stresses are initially elastic. However, cycling hardening occurs in Alloy 617, which can have both beneficial and disadvantageous effects on the cyclic life of the parts subject to high temperatures. For more detailed reading on high temperature fatigue, see appendix B of this report. Data from the literature are taken to obtain the fatigue behaviour. Most data are based on laboratory tests, which are performed with zero mean stress and do not consider hold-times. Hold-times are considered in the creep-fatigue part later in this report.

2.5.1 Fatigue curve construction

Fatigue can be subdivided into a high-cycle fatigue (HCF) part and a low-cycle fatigue (LCF) part. The HCF part is described by the elastic strain-life curve expressed by the Basquin relation [51,52]:

$$\sigma_a = \sigma_f' \left(2N_f \right)^{b_f} \qquad , \tag{2.5A}$$

in which σ_a is the stress amplitude, σ'_f is the fatigue strength coefficient, $2N_f$ is the number of load reversals to failure and b_f is the fatigue strength exponent. LCF is described by the plastic strain-life curve and can be calculated through the Coffin-Manson relation that uses the plastic strain amplitude $(\varepsilon_{a,p})$ [51,53,54]:

$$\varepsilon_{a,p} = \varepsilon_f' \left(2N_f \right)^{c_f} \qquad , \tag{2.5B}$$

where ε'_{f} is the fatigue ductility coefficient and c_{f} is the fatigue ductility exponent. Commonly, fatigue damage is caused by cyclic strains and therefore the strain amplitude (ε_{a}) is the essential loading parameter for correlating cyclic life [51,55]:

$$\varepsilon_a = \frac{\sigma'_f}{E} \left(2N_f \right)^{b_f} + \varepsilon'_f \left(2N_f \right)^{c_f} \quad , \tag{2.5C}$$

with E is the elastic (Young's) modulus at the temperature of evaluation.

Figure 2.21 shows the development and dominance of these two mechanisms.



Figure 2.21: Elastic and plastic strain amplitude versus life graphs and their superposition [52,56], where $\Delta \varepsilon/2 = \varepsilon_a$.

The data for fatigue examination are gathered from smooth surface test samples and under laboratory conditions. To convert the best-fit curve through these data points to a fatigue design curve for the use in an industrial environment, three conservative factors have to be incorporated in the tests results [58]. These factors are:

- 2.0 Scatter of data (difference between minimum and mean values).
- 2.5 Grain size effect (i.e. differences in grain size).
- 4.0 Environmental conditions (e.g. atmosphere and surface finish).

The 'scatter of data' factor includes effects of differences in chemical composition, cold work, and heat treatment. The 'grain size effect' factor tackles grain size differences due to manufacturing, chemical composition (that dictates the number of nuclei), through thickness microstructure variation (i.e. the grain size at the surface deviates from the size in the core/centre) and grain growth due to unforeseen service process influences (e.g. extra heat). This effect is important, since commonly finer grains means more grain boundaries which in turn hinder fatigue. In the 'environmental conditions' factor, aspects are included like surface finish, atmosphere, small notches/scratches and oxidation. All three factors are relevant to the number of cycles to failure and these are multiplied with each other, so a total factor of 20 is obtained. On the stress amplitude (note: different than the strain amplitude as is explained later in this paragraph) only a factor 2.0 for the 'scatter of data' is applied. Not both factors, i.e. 2.0 on the stress amplitude and 20 on cycles to failure, have to be applied, but only the factor that creates the most conservative fatigue design curve. This causes a 'dent' in the fatigue design curve at the intersection of the curves belonging to these two factors (see for example the design curve in figure 2.22 in the next sub-paragraph).

2.5.2 Fatigue curve for temperatures lower than 425 °C

Based on fatigue data produced and conservative factors applied, a hypothetical fatigue design curve has been proposed, which shall be located about a factor 1.3 in stress amplitude below the fatigue design curve I-9.5M from the ASME III, appendix I for numbers of cycles to failure fewer than 10⁵ (shown in figure 2.22) [56]. Subsequently, from 10⁵ cycles up, this factor decreases evenly with the number of cycles to failure until the curves coincide at 10⁶ cycles.



Figure 2.22: Hypothetical design curve for Alloy 617 for below 425 °C [56].

From 10^6 cycles and more the Alloy 617 curve shows the same trend as fatigue design curve I-9.5M. The I-9.5M fatigue design curve, as shown in figure 2.23, is meant for nickel-chromium-molybdenumiron alloys, and due to the Alloy 617's nickel-base matrix containing a considerate number of alloying elements including chromium and molybdenum, it is found that the fatigue behaviour is quite similar [56]. Other fatigue data is not available for Alloy 617 in the temperature range below 425 °C and for that reason the hypothetical design curve is used.



FIG. 1-9.5M DESIGN FATIGUE CURVES FOR NICKEL-CHROMIUM-MOLYBDENUM-IRON ALLOYS

Figure 2.23: Design fatigue curve for Ni-Cr-Mo-Fe alloys below 425 °C [17], where $N = N_f$.

Commonly a best-fit curve for the test data is used in which the stress amplitude (σ_a) is equated as function of the number of cycles to failure (N_f) [57]:

$$\sigma_a = \frac{E}{4\sqrt{N_f}} \log(A_f) + B_f \quad , \tag{2.5D}$$

in which A_f is the adjustment constant for the best-fit curve and B_f is the endurance limit (i.e. stress amplitude below which no rupture occurs) at 10⁷ cycles, which is 173 MPa based on figure 2.23. Equation (2.5D) is re-arranged to obtain the number of cycles (N_f) as function of the stress amplitude:

$$N_f = \left(\frac{E \cdot \log(A_f)}{4(\sigma_a - B_f)}\right)^2 \qquad (2.5E)$$

Unfortunately, the fatigue design curve for Alloy 617, as it is supposed to evolve from the fatigue design curve I-9.5M as shown in figure 2.23, cannot be described completely using equation (2.5E). Therefore, the curve has been divided into five parts. For stresses larger than 318.3 MPa (equivalent to $2.1 \cdot 10^4$ cycles), the value of A_f can be described with:

$$A_f = -0.22819(\sigma_a)^4 + 2.8753(\sigma_a)^3 - 12.767(\sigma_a)^2 + 23.044 \cdot \sigma_a - 11.586$$
(2.5F)

With stress amplitudes between 194.0 and 318.3 MPa (equivalent to $2.1 \cdot 10^4 - 1.0 \cdot 10^6$ cycles), the value of A_f is constant at 2.63.

Between the stress amplitudes 165.6 - 194.0 MPa (i.e. $1.0 \cdot 10^6 - 2.5 \cdot 10^8$ cycles), equation (2.5E) cannot be used, because the stress amplitude (σ_a) reaches the value of the endurance limit (B_f), causing equation (2.5E) to behave asymptotic. Therefore, best-fit plots have been made for the stress amplitude range of 173.0 MPa – 194.0 MPa (equivalent to $1.0 \cdot 10^6 - 1.0 \cdot 10^7$ cycles):

$$N_f = 10^{\left(-5.5607 \cdot 10^{-5} (\sigma_a)^3 + 3.1876 \cdot 10^{-2} (\sigma_a)^2 - 6.1227 \cdot \Delta\sigma + 4.0013 \cdot 10^2\right)} , \qquad (2.5G)$$

and for the stress amplitude range of 165.6 – 173.0 MPa (equivalent to $1.0 \cdot 10^7 - 2.5 \cdot 10^8$ cycles):

$$N_f = 10^{(1.9725 \cdot 10^{-2}(\sigma_a)^2 - 6.8661 \cdot \sigma_a + 6.0449 \cdot 10^2)}$$
(2.5H)

For stress amplitudes lower than 165.6 MPa, a linear relation is observed between the logarithmic cycles to failure value as a function of the stress amplitude:

$$N_f = 10^{(-\sigma_a + 174)} (2.51)$$

This linear relation is representing the elastic Basquin part of the fatigue equation, because plastic deformation is limited in this part of the fatigue curve.

2.5.2.1 Fatigue data in the temperature range 425 - 538 °C

Fatigue data in the temperature range 425 - 538 °C are not available. The reason for this lack of data is given in the discussion section in chapter 6 of this report. In short, the application of Alloy 617 in this temperature range is not cost efficient.

2.5.3 Fatigue curve for temperatures above 538 °C

For the evaluation of fatigue at temperatures above 538 °C, the strain – cycle life assessment method is used. Six parameters are required for this evaluation:

- Strength coefficient σ'_f
- Strength exponent b_f
- Ductility coefficient ε'_f
- Ductility exponent c_f
- Cyclic strength coefficient K'
- Cyclic strain hardening exponent n'

The relation between the first four fatigue input parameters is a combination of the Basquin (for the elastic part) and Coffin-Manson (for the plastic part) equations as shown in figure 2.21 and given as equation (2.5C) as shown in paragraph 2.5.1. The four fatigue parameters in equation (2.5C) (i.e. σ'_f , b_f , ε'_f and c_f) are constants, and the elastic modulus (*E*) is a temperature-dependent parameter. The elastic modulus decreases with an increase in temperature, which in turn increases the $\frac{\sigma'_f}{E}$ term. When regarding a fixed strain amplitude and a stress below the yield strength, the portion of elastic strain becomes larger with an increase in temperature. This effect is shown in figure 2.24. At a temperature of 600 °C, the yield strength is 159.3 MPa and the elastic modulus is 173.3 GPa, and as result the elastic strain at yield strength is 9.19 $\cdot 10^{-4}$ by using Hooke's law. By increasing the temperature to 800 °C, the yield strength does not change significantly, i.e. becomes 157.2 MPa which is a reduction of about

1.3%, while the elastic modulus diminishes by almost 10% to 157.3 GPa. At 800 °C, the elastic strain at the yield strength as calculated by Hooke's law is $1.00 \cdot 10^{-3}$, which is about 10% larger than the elastic strain at 600 °C for approximately the same stress (i.e. yield strength).



At stresses lower than the yield strength, fatigue damage develops due to microscale creep and/or plastic deformation due to cyclic hardening.

Figure 2.24: Effect of temperature on the elastic modulus, yield strength and elastic strain.

An additional temperature influence is the faster conversion of elastic strain into plastic strain due to faster creep, more softening of the material and less cyclic hardening with the number of cycles, which changes the relative ratio between the elastic and plastic strain. This is shown by the diminished slope of the Basquin part of equation (2.5C), i.e. b_f , with an increase in temperature. This diminished slope causes σ'_f to change as well.

The cyclic strain hardening exponent (n') and cyclic strength coefficient (K') can be determined in several manners, though the following relations for these factors are most common [59]:

$$n' = \frac{b_f}{c_f} , \qquad (2.5J) \text{ and}$$
$$K' = \frac{\sigma'_f}{\varepsilon'_f^{c_f}} = \frac{\sigma'_f}{\varepsilon'_f^{n'}} . \qquad (2.5K)$$

The fatigue parameters for several temperature ranges are shown in table 2F.

Т	σ'_f	b _f	ε_{f}	C _f	K'	n'
[°C]	[Pa]	[-]	[-]	[-]	[Pa]	[-]
538-704	1.330E+09	-0.11	0.68	-0.76	1.406E+09	0.145
704-871	1.160E+09	-0.09	0.68	-0.76	1.214E+09	0.118
871-982	1.085E+09	-0.08	0.68	-0.76	1.130E+09	0.105

Table 2F: Values of the fatigue parameters in different temperature ranges [60].

The effect of the cycling hardening influences the relation between the stress amplitude (σ_a) and the strain amplitude (ε_a) [59]:

$$\varepsilon_a = \frac{\sigma_a}{E} + \frac{1}{2} \left(\frac{\sigma_a}{K'} \right)^{\frac{1}{n'}}$$
(2.5L)

The temperature dependency of the strain amplitude in figure 2.21 is covered in the elastic modulus and fatigue parameters. Therefore, the fatigue curve for a certain temperature can be constructed around one single point in the graph. The coordinates of a point on the fatigue curve that can be determined accurately is the transition fatigue life in number of reversals at which the elastic and plastic strains are equal. This is referred to as the transition point. The transition fatigue life in number of reversals ($2N_t$) is found by [59]:

$$2N_t = \left\{ \frac{E\varepsilon_f'}{\sigma_f'} \right\}^{\overline{(b_f - c_f)}}$$
(2.5M)

Using equation (2.5C) and (2.5M), the total strain amplitude at the transition fatigue life (ε_t) can be calculated as:

$$\varepsilon_t = \frac{\sigma_f}{E} (2N_t)^{b_f} + \varepsilon_f' (2N_t)^{c_f} \qquad (2.5N)$$

From the point of intersection of the elastic strain amplitude line and the plastic strain amplitude line (i.e. $\frac{1}{2}\varepsilon_t$), the change in the elastic strain amplitude ($\Delta \varepsilon_e$) behaves in the following manner:

$$\log\left(\Delta\varepsilon_e\right) = b_f \cdot \log\left(\Delta 2N\right) \quad , \tag{2.50}$$

in which $\Delta 2N$ is the change in number of load reversals and the change in the plastic strain amplitude $(\Delta \varepsilon_p)$ can be described with:

$$\log(\Delta \varepsilon_p) = c_f \cdot \log(\Delta 2N) \qquad (2.5P)$$

The change in the total strain amplitude ($\Delta \varepsilon_a$) is the sum of equations (2.50) and (2.5P) and also the deduction of the total strain amplitude at the transition point (ε_t) from the total strain amplitude (ε_a):

$$\Delta \varepsilon_a = \Delta \varepsilon_e + \Delta \varepsilon_p = \varepsilon_a - \varepsilon_t \qquad (2.5Q)$$

The mathematics for determining the number of cycles to failure is described and explained in appendix D, paragraph D4.

2.5.4 Cycle count methods

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In case of non-proportional loading (i.e. subsequent cycles do no reach the same stress or strain value or they are different cycle types), the max-min cycle counting method is most suitable to determine the total time representing each stress/strain range and/or cycle type. In this method, all expected loads are rearranged [16-18,61]. The first cycles evaluated are all largest load fluctuations, then the second largest, etc., until the smallest load ranges have been incorporated. This leads to the sum of fatigue damages caused by fluctuating loads:

- a. STEP 1 From the load histogram determine the sequence of peaks and valleys.
- b. **STEP 2** Calculate the stress components (σ_{ij}) resulting from the loading in each time-step. The stress components have to be put into the same global coordinate system. In case of local

discontinuities, the stress analysis has to include the peak stresses at those locations. The controlling stress for fatigue evaluation is the effective total equivalent stress amplitude, which is half the effective total equivalent stress range, which is derived from the sum of the primary and thermal stresses.

- c. **STEP 3** Scan (local) time points and delete the time points at which none of the stress components show reversals (i.e. no peaks or valleys).
- d. **STEP 4** Determine the time point with the highest peak and lowest valley. This time point shall be indicated as ${}^{m}t$ and the corresponding stress components as ${}^{m}\sigma_{ij}$.
- e. **STEP 5** Determine the stress component range between the peak and valley of ${}^{m}t$. The next time point is ${}^{n}t$ with stress components ${}^{n}\sigma_{ij}$. Then calculate the stress component ranges and the Von Mises equivalent stress range between time points ${}^{m}t$ and ${}^{n}t$:

$${}^{mn}\sigma_{ij} = {}^{m}\sigma_{ij} - {}^{n}\sigma_{ij}$$

$${}^{mn}\Delta S_{range} = \frac{1}{\sqrt{2}} \sqrt{ \frac{\binom{mn}{\Delta\sigma_{11} - mn}{\Delta\sigma_{22}}^2 + \binom{mn}{\Delta\sigma_{22} - mn}{\Delta\sigma_{33}}^2 + \binom{mn}{\Delta\sigma_{33} - mn}{\Delta\sigma_{11}}^2 + 6\binom{mn}{\sigma_{12}^2 + mn}{\sigma_{23}^2 + mn}{\sigma_{13}^2}}_{(2.5S)}$$

(2.5R)

- f. **STEP 6** Repeat STEP 5 for each time point in the stress histogram.
- g. **STEP 7** Obtain the maximum equivalent Von Mises stress in STEP 5. Record the time points ${}^{m}t$ and ${}^{n}t$ which define the start and end points of the cycle.
- h. **STEP 8** Determine the event(s) to which ${}^{m}t$ and ${}^{n}t$ belong to and record their respective specified number of repetitions as ${}^{m}N$ and ${}^{n}N$.
- i. **STEP 9** Determine the number of repetitions of the cycle.
 - 1. If ${}^{m}N < {}^{n}N$: delete ${}^{m}t$ from those considered in STEP 4 and reduce the number of repetitions at ${}^{n}t$ from ${}^{n}N$ to (${}^{n}N {}^{m}N$).
 - 2. If ${}^{m}N > {}^{n}N$: delete ${}^{n}t$ from those considered in STEP 4 and reduce the number of repetitions at ${}^{m}t$ from ${}^{m}N$ to $({}^{m}N {}^{n}N)$.
 - 3. If ${}^{m}N = {}^{n}N$: delete both ${}^{m}t$ and ${}^{n}t$ from those considered in STEP 4.
- j. STEP 10 Repeat STEPs 4 to 10 until all time points with stress reversals have been addressed.
- k. **STEP 11** Use the data recorded for the counted cycles to perform the fatigue assessment. An elastic-plastic fatigue assessment is required in case ${}^{mn}\Delta S_{range}$ exceeds the yield point of the cyclic stress range-strain range curve of the material.

These steps are automatically calculated by the finite element software program.

The total damage that is caused by fatigue is determined as follows:

- a. **STEP 1** Determination of the number of cycles to failure $(N_{f,k})$ of the kth cycle.
- b. **STEP 2** Computation of the fatigue damage parameter for the kth cycle $(D_{f,k})$ with the actual number of repetitions of the kth cycle (n_k) , as:

$$D_{f,k} = \frac{n_k}{N_{f,k}} \qquad (2.5T)$$

- c. **STEP 3** Repetition of STEP 2 for all stress ranges identified in the cycle counting until the total number of cycles (*M*) is reached.
- d. **STEP 4** Calculation of the accumulated fatigue damage by summation of the values computed at STEP 3. The protection against failure from cyclic loading is maintained in case:

$$\sum_{k=1}^{M} D_{f,k} \le 1.0$$
 (2.5U)

2.6 Thermal ratchetting and thermomechanical fatigue

Thermal (cyclic) stresses develop due to a thermal gradient over the wall thickness of a pressurised component and the thermal stress range (S_Q) can be calculated based on the coefficient of thermal expansion (α), the elastic modulus (E) and the temperature difference (ΔT) between the inner and outer surface of the pressurised component:

$$S_O = \alpha \cdot E \cdot \Delta T$$
 . (2.6A)

Ratchetting is a combination of a constant primary stress and a cyclic thermal stress that causes plastic strains to progress and accumulate during each subsequent cycle. Ratchetting can be avoided, which basically implies that plastic strains are prevented to accumulate. This can be accomplished by limiting the cyclic thermal stress which is caused by a thermal gradient over the wall thickness of the pressurised component. First, the ratio (X) of the primary membrane stress (σ_p) to the minimum yield strength at the maximum temperature of the cycle ($\sigma_{y,Tmax}$) is determined [16-18]:

$$X = \frac{\sigma_p}{\sigma_{y,Tmax}}$$
 (2.6B)

For thin-walled components, like boiler tubes, the stress distributions through the wall thickness are considered linear. The thermal stress range limit (S_O) to avoid ratchetting is set to [16-18]:

$$S_Q = \sigma_{y,Tmax} \left(\frac{1}{X}\right)$$
, for $0 < X < 0.5$ (2.6C) and
 $S_Q = 4\sigma_{y,Tmax}(1-X)$, for $0.5 \le X \le 1.0$. (2.6D)

Equations (2.6C) and (2.6D) have their origin in the Bree diagram [62,63] as shown in figure 2.25, left figure. In this figure, the primary constant stress is shown as a ratio of the yield strength along the horizontal axis and the cyclic thermal stress (in the figure indicated with σ_t) is given as a multiplication of the yield strength along the vertical axis. Equation (2.6C) is the equation for the line that separates the P (plastic shakedown) and R (ratchetting) zone. Equation (2.6D) is the equation for the line between the S (elastic shakedown) and R (ratchetting) zones. Elastic shakedown is the mechanism in which a small plastic strain is developed during the first cycle and during subsequent cycles only elastic strains are observed (see also the right figure in figure 2.25). In the region underneath these two lines, ratchetting (i.e. zone R) is avoided. A parabolic variation of stresses through the wall thickness can be used for thick-walled components, though this is not considered in this research since thick-walled components are not of interest at the moment.

Besides the equations to avoid ratchetting, it is advisable to limit the thermal stress range to a maximum of twice the yield strength to avoid cyclic plastic deformation during each cycle, which leads to plastic shakedown, i.e. low-cycle fatigue. Therefore, an additional limit is given for the thermal stress, which is shown in the left figure of figure 2.25 by the line between the plastic shakedown zone (P) and the elastic shakedown zone (S):

$$S_O = 2\sigma_{y,Tmax} (2.6E)$$

Since both equations (2.6C) and (2.6E) are applicable for 0 < X < 0.5, the later equation is used because it is the most limiting equation for this range of *X*.



Figure 2.25: left figure: Bree diagram for elastic cycling (E), elastic shakedown (S), plastic shakedown (P; i.e. low-cycle fatigue) and ratchetting (R) [62]. Right figure: Detail of lower thermal stress part of Bree diagram [63].

Thermal ratchetting or low-cycle fatigue is not expected to occur when the secondary thermal stress range (ΔQ) is smaller than/equal to the thermal stress range limit (S_O):

$$\Delta Q \le S_Q \qquad . \tag{2.6F}$$

Equations (2.6D) and (2.6E) show that the thermal stress range limit (S_Q) depends on the minimum yield strength at the highest temperature ($\sigma_{y,Tmax}$). In case the primary stress (σ_p) is larger than half the yield strength at temperature, this primary stress has to be considered as well. The maximum allowable thermal stress range (ΔQ_{max}) is therefore:

$$\Delta Q_{max} = 2\sigma_{y,Tmax} , \qquad \text{for } 0 < X < 0.5 , \qquad (2.6G) \text{ and}$$

$$\Delta Q_{max} = 4(\sigma_{y,Tmax} - \sigma_p) , \qquad \text{for } 0.5 \le X \le 1.0 . \qquad (2.6H)$$

More basic reading on ratchetting and shakedown can be found in Appendix B. A detailed description of the Bree diagram with its zones and equations is given in Appendix E.

2.7 Creep-Fatigue

The major damage mechanism for materials in service in high-temperature processes might be creepfatigue (CF) damage, which is caused by cycling due to start-up and shutdown or power transients combined with sustained loading [3,64]. The CF interaction affects the mechanical properties resulting in shorter lifetimes compared to the sum of creep damage and fatigue damage incurred separately [3,64-67]. This is applicable irrespective of the hold position, but the damage is more severe for load holds in the tension direction [3].

2.7.1 Creep-fatigue interaction

Creep-fatigue interaction is best illustrated by dividing the Alloy 617 material into separate alloy phases. Both nickel matrix (γ) and Ni₃Al-precipitates (γ') have a face-centred cubic (FCC) lattice, which leads to a set of twelve slip systems that carry the plastic deformation [68]. The two phases share a coherent interface with a small misfit due to a slight difference in lattice constants. To represent the microstructure, a unit cell of 16 regions can be considered (see figure 2.26) one precipitate (γ') region, three matrix (γ) regions with different orientations that act as channels, and six pairs of γ/γ' interface (I) regions (so twelve interface regions in total) [68,69]. The regions representing the matrix and the precipitate incorporate the behaviour of the pure phase. The interface regions deal with the short-range processes and effects that occur at the γ/γ' interface, like interaction between the phases and dislocations induced by back stress.



Figure 2.26: Multiphase unit cell for calculating the mechanical material response [69].

2.7.1.1 Creep damage accumulation

Creep damage accumulation is often caused by the formation and growth of micro-voids at the γ/γ' interfaces [70]. At these interfaces, the internal stress is higher which attracts atom vacancies from the matrix resulting in an accumulation of vacancies. When sufficient atom vacancies are gathered, a micro-void is formed. These micro-voids elongate and eventually coalesce prior failure. Another important creep mechanism in nickel alloys, like Alloy 617, is rafting, especially for temperatures over 850 °C. Under a certain amount of stress, the initially cuboidal precipitates grow into elongated plates [71-73]. The direction in which the precipitates grow is dependent on the applied load direction and the sign of lattice misfit as shown in figure 2.27 [74,75]. A positive misfit means that the interatomic distances inside the precipitate are larger than those inside the matrix. This causes the morphology of the matrix (i.e. the γ -channels) to change. Since the matrix suffers the largest part of the plastic deformation, rafting has a significant effect on the mechanical response of the material. These creep mechanisms have already been incorporated in the equations determined in paragraph 2.3.



Figure 2.27: Rafting of γ -precipitates in a nickel-based alloy [74].

2.7.1.2 Fatigue damage accumulation

Lattice planes slip along each other due to dislocation motion, which results in deformation that is irreversible. The irreversibility of this slip is the cause for fatigue damage. On a microscopic level, during forward slip, the dislocation line is forced in between two precipitates when the effective shear stress τ_{eff} (i.e. the net effect of both externally applied and internal stress on the slip plane in the slip direction) exceeds the Orowan threshold (τ_{Or}), which is the shear stress required for mobile dislocations to bypass the γ' precipitates by Orowan bowing [37,69]:

$$\tau_{eff} \ge \tau_{Or} = \alpha_{Or} \frac{\mu b}{L_S} \qquad , \qquad (2.7A)$$

in which α_{Or} is a dimensionless constant related to the dislocation core radius (α_{Or} = 0.85 [69]), μ is the shear modulus, b is the Burgers vector, and L_s is the spacing between the precipitates. Figure 2.28 shows forward slip shown as line (1) moving towards the situation of line (2). Ideally, these dislocations glide back in the original position (1) upon removal of the load and between other precipitates in the opposite direction upon a reversal of the load (i.e. position (3)).



Figure 2.28: Schematic of dislocation line movement on a {1 1 1} plane in the matrix phase. Ideal case where a dislocation line (1) is forced between two precipitates (2) by a stress. Stress removal will cause the dislocation line to move back to its original position (1). An opposite stress pushes the line between two other precipitates (3) [76].

This ideal situation does not occur, since part of the dislocations become immobilised. Due to the dislocation line movement, dislocation loop trailing segments are deposited onto the γ/γ' interface as is shown in figure 2.29, left figure. Backward slip occurs in case the effective stress diminishes to a value below the Orowan threshold ($\tau_{eff} < \tau_{Or}$) [68]. An unhindered and mobilised backward glide of the dislocation loop should develop as shown in the middle figure of figure 2.29. In practice, the backward slip is hindered by the deposited dislocation loop trailing segments, which make the velocity of dislocations moving backward different than their forward velocity as shown in the right figure of figure 2.29.



Figure 2.29: Dislocation line forced into the γ -channel between two γ' precipitates; left figure: segments of the line are deposited onto the γ/γ' interfaces; middle figure: on reduction of the applied stress the loop moves backward; right figure: the interfacial segments are immobilised and backward slip is only possible by bowing in the opposite direction resulting in the deposit of new segments on the interface [68].

Mechanisms hindering the backward motion [68]:

- Bulk material loop immobilisation, i.e. dislocation loops are immobilised by reactions with other loops.
- γ/γ' interfacial segment immobilisation, i.e. the segments are immobilised preventing the complete dislocation loop from moving back (as is shown in figure 2.29, right figure). This can be due to:
 - o interfacial segments react with segments of other slip systems and form networks;
 - o segments affect the misfit locally, which results in coherency loss;
 - \circ interfacial segments leave the original slip plane by climb or cross slip.

In these cases, other slip planes take over and damage accumulates with the number of cycles, which eventually results in the initiation of cracks. At the surface of the material, the irreversibility is caused by the exiting of dislocations at the free surface [77]. In the bulk material, a constitutive approach is

chosen to describe the amount of reverse slip that occurs after a load reversal [76]. The reverse slip process in the bulk material is partly influenced by internal back stresses and therefore a distinction has to be made between forward and backward slip. The immobilisation process is considered a combined isotropic and kinematic hardening effect. Each immobilised interfacial segment increases the resistance against dislocation motion in either direction (i.e. isotropic hardening). Therewith, the deposited dislocation loop trailing segments contribute to the internal back stress, in which the sign of its contribution depends on the direction of the dislocation movement [69]. This contribution to the internal back stress provides a kinematic effect. Therefore, the quantification of the amount of immobilisation (i.e. hindering of dislocations) is important.

On a macroscopic scale, the fatigue behaviour is covered in paragraph 2.5, where the fatigue in number of cycles to failure is related to the strain amplitude.

2.7.1.3 Creep-Fatigue mechanism

The interaction between creep and fatigue enhances damage accumulation. The creep damage mechanism is the formation, growth and coalescence of voids and the damage mechanism of fatigue is the immobilisation of dislocation segments resulting in slip irreversibility. Voids caused by creep introduce additional free surfaces inside the material, which create dislocation annihilation sites and consequently enhance slip irreversibility (i.e. fatigue). Fatigue cyclic loading induces pile-ups of immobilised dislocations, which in turn are potential sites for creep void formation. For these reasons, the presence of one damage mechanism affects the amount and evolution rate of the other damage mechanism.

2.7.2 Hold-time influence

During the hold-time at a fixed strain (see figure 2.30), creep strain develops in a process called stress relaxation. This process is described in paragraph 2.3.4 as well. The creep strain accumulation causes more internal damage, which in turn leads to an increased chance of crack initiation and accelerates crack growth, resulting in a reduced fatigue cycle life [65].



Figure 2.30: Strain-controlled fatigue cycle with hold periods (a) imposed strain over time; (b) stress response over time [79].

The introduction of tensile hold-times (i.e. minimum hold-time tested is 180 seconds) during cycling diminishes the total cycle life immediately with roughly factor 2 for all strain ranges [6,66,78] in comparison to the cycle life of continuous-cycle fatigue, and going to a factor 3 for longer hold-times [64]. Above a certain tensile hold-time (depending on the conditions), the number of cycles to failure is not influenced anymore by the tensile hold-time [64]. This effect is more pronounced for low total strain ranges than for high total strain ranges (i.e. \geq 1.0%) [6,66,78].

As can be seen from the example shown in figure 2.31, fatigue and CF midlife hysteresis loops make evident that the introduction of a tensile hold-time increases the amount of inelastic (i.e. creep) strain, which in turn results in a measured stress level decay, i.e. stress relaxation, over the strain-controlled hold, as is also shown in figure 2.32a [6,64]. A constant reduced stress is obtained after a certain period from the hold-time initiation. In figure 2.32b, the peak tensile stresses and the approximated relaxation stresses are shown for a relatively high total strain range. Concluded is that the relaxed stresses are similar for all strain ranges for tensile hold-times between 180 and 9,000 seconds [6,64] and that cyclic life is reduced significantly.



Figure 2.31: Hysteresis loop for 0.3% total strain range and no tensile hold-time and 600 s hold-time [64].



Figure 2.32: a) Rapid stress relaxation during the tensile hold, CF test at 0.3% total strain range; b) Peak tensile (dots) and relaxed (circles) stresses versus cycles to failure at 1.0% total strain range [64].

2.7.3 Damage accumulation model

A creep-fatigue damage accumulation model should cover a broad range of loading conditions in a computationally efficient manner. The model has to account for load histories that are more complex than standard constant amplitude/constant temperature load sequences. Also, the model has to contain a time-dependent component which covers diffusion and dislocation movement (i.e. creep segment), a cycling waveform and cycling period (i.e. fatigue segment), and a thermo-mechanical component, which describes the deformation caused by thermal expansion/contraction (i.e. shakedown, ratchetting). These different damage segments have been equated in the previous paragraphs of this chapter.

2.7.3.1 Uncoupled visco-plastic model

An efficient damage rule model is uncoupled (i.e. post-processing based), so the analysis of one or a limited number of representative cycles is sufficient to determine the time or the number of cycles to failure. A coupled model would require the complete load sequence to be analysed, since the damage accumulation leads to a continuously changing material response. Experience with nickel alloys (NA) has learned that NAs commonly do not show considerable cyclic softening prior failure [80]. This means that damage evolution has a limited effect on the constitutive behaviour and that application of an uncoupled damage model is valid [68].

Visco-plastic material models combine classical fatigue methods, like that of Coffin-Manson (as described in paragraph 2.5) or Palmgren-Miner [81,82], and creep-life methods, which are based on Norton's creep law (same as power-law creep; see paragraph 2.3) [70], and transform these into time-incremental damage rules [83,84].

2.7.3.2 Creep-fatigue interaction part of the model

Classical rules, like the Palmgren-Miner rule, suggest that the contributions of the separate damage mechanisms are independent of each other (i.e. do not interact), resulting in linear damage accumulation (dashed black line in figure 2.33, constructed for material CMSX-4 (Nickel based single crystal; Ni-9Co-6.5Cr-6W-5.6Al) and illustrated as an example). The limiting cases of such linear damage accumulation are the pure creep tests (i.e. no cycling) and the continuous cyclic tests (i.e. no hold-times to avoid creep development). Actual experimental values from LCF tests, which contain both a creep and a fatigue damage part, give a non-linear trend as is shown as an example for CMSX-4 in figure 2.33 by the blue dotted line. This is a clear indication that creep and fatigue damages interact and enhance each other.

The creep-fatigue interaction can be taken into account to determine the total damage accumulation by [68]:

,

$$D_{tot} = D_c + D_f + A_{int} \frac{D_c \cdot D_f}{D_c + D_f}$$

(2.7B)

in which:

 D_{tot} = Total damage accumulation.

 D_c = Creep damage parameter.

 D_f = Fatigue damage parameter.

 A_{int} = Interaction coefficient.



Figure 2.33: Creep-fatigue interaction of CMSX-4 at 850 °C [68].

As indicated in figure 2.33 by the green line, the damage accumulation model can be made to agree with the experimental trend (i.e. the blue dotted line) for the nickel based single crystal CMSX-4.

2.8 Design code requirements, restrictions and limitations

The most frequently used design code for the design of pressurised components inside a steam boiler, is the ASME (i.e. American Society of Mechanical Engineers) code [13,17,18]. In the ASME code sequence, ASME II deals with material standards and inside this code part, section D lists the maximum allowable stresses for the materials in temperature intervals of 25 °C. The maximum allowable stress acting on a boiler component at a certain temperature. In this paragraph, a summary is given regarding the requirements, restrictions and limits enforced upon by the ASME design code. An extended description can be found in Appendix F.

Together with the API (American Petroleum Institute), the ASME has also developed Fitness-For-Service (FFS) assessment procedures with the purpose to evaluate pressurised components containing/accumulating damage and these procedures have been described in the API 579-1/ASME FFS-1 [16]. More detailed information about the FFS assessment procedures is given in Appendix C.

2.8.1 Comparison of the maximum allowable stresses

The ASME II Code, section D, gives the values of the maximum allowable stress for each temperature interval of 25 °C. The sources for these values are not given by the ASME code, except that they are the lowest value of the stresses that cause a certain amount of damage by the damage mechanisms as described in paragraphs 2.2 to 2.4 of this chapter. This lowest value has a conservative factor included and the damage mechanism leading to the lowest stress value is considered dominant. More detailed information about the maximum allowable stress from the ASME design code is added to Appendix F, paragraph F1. By adapting the equations in paragraphs 2.2 to 2.4, similar values should be obtained as given by Section D of the ASME II code [13].

2.8.1.1 Maximum allowable stress based on yield strength

One of the maximum allowable stresses based on the yield strength is maximised to 160 MPa, which is $\frac{2}{3}$ of the yield strength at room temperature (i.e. 240 MPa for Alloy 617). This maximum is set because the hydrostatic test of the pressurised components is performed at 1.5x the design pressure at room temperature to ensure the integrity of these components. To avoid plastic deformation during the hydrostatic test, the design stress (i.e. the maximum allowable stress) is determined at the yield strength at room temperature divided by a factor 1.5.

For some materials, the ASME design code gives the option to use either $\frac{2}{3}$ or 90% of the yield strength at temperature. Alloy 617 is one of these materials. The percentage of 90 may be used instead of factor $\frac{2}{3}$ when a slightly higher deformation is in itself not objectionable. This is applicable to Alloy 617 boiler components, because Alloy 617 has a large ability for strain hardening since it contains a FCC structure with a small yield-strength-to-ultimate-tensile-strength ratio and because boiler components commonly do not have complex geometries, i.e. no large stress concentrations.

When the maximum allowable stresses from the ASME II code, section D, are plotted for temperatures below 640 °C, the trend-line (indicated by the red asterisks in figure 2.4 in paragraph 2.2.2) is quite similar to that shown in figure 2.5, only all values from the ASME II-D, are about 8 MPa lower than determined in paragraph 2.2.2 from the 95% lower prediction bound of the raw data. Although the ASME code does not indicate how these values are determined, the most likely reason is that the maximum allowable stress from the ASME is determined at 90% of the yield strength at temperature, which is lower than the 95% lower predication bound of the raw data in figure 2.4. In the calculations, this difference is compensated for by lowering the a_1 of the normalised yield strength in equation (2.2A) and table 2C by 0.0333 (i.e. 8 MPa/240 MPa), in which the dominator is the value of minimum yield strength at room temperature. Also, the plot through the ASME maximum allowable values seems to be slightly steeper than the plot created from the normalised data. This difference may be caused by the use of other datasets by the ASME code and this is compensated for by increasing b_1 of the normalised yield strength with a value of 0.03. The adjusted equation for the maximum allowable stress based on the yield strength ($\sigma_{YS,allow}$) becomes:

$$\sigma_{YS,allow.} = 0.9 * 240(0.65123 + 0.39097 \exp(-5.7516 \cdot 10^{-3}T))$$
[MPa]
(2.8A)

with T the temperature in degrees Celsius. This adjusted equation for the minimum yield strength is used in the remainder of the report, since the values calculated with this equation describe the maximum allowable stress values according to the ASME II and therefore limit the applied stresses.

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2.8.1.2 Creep stress to rupture

ASME code Section II, Part D, Appendix 1, states that the maximum allowable stress based on the creep rupture strength is the stress at which rupture occurs after 100,000 hours of exposure to a certain temperature. To obtain this stress, the Larson-Miller Parameter (*LMP*) has to be determined at a time to rupture (t_R) of 100,000 hours, and with the Log t_R = Log (100,000) = 5, equation 2.4B can be rewritten as (with T_K in Kelvin):

$$LMP = T_K(\log t_R + 20)) = 25T_K$$
 (2.8B)

When plotting the maximum allowable stress values from ASME II code, section D, for above 640 °C, a trend is observed similar to that of the creep rupture strength as determined with the equations from paragraph 2.4, except that the slope of the data plot in the ASME Larson-Miller diagram is less steep. This is probably caused by the different data set(s) used by the ASME design code. Adjusting the equations from paragraph 2.4, expressing the maximum allowable stress based on the creep rupture strength ($\sigma_{R,allow}$) as function of the temperature T_K in Kelvin and taking a conservative factor of 20% into account, which is enforced upon by the ASME (i.e. $\sigma_{R,allow}$ is 80% of the minimum creep rupture strength ($\sigma_{R,min}$)), results in:

$$\sigma_{R,allow.} = 0.8 \sigma_{R,min.} = 8.2081 \cdot 10^5 \cdot \exp(-9.4801 \cdot 10^{-3} T_K).$$
 (2.8C)

This equation is used in the remainder of the report for the determination of the creep rupture strength for life-times smaller than or equal to 100,000 hours. For life-times larger than 100,000 hours, this ASME limitation has to be more conservative, since for design approval the conservatism of the ASME has to be maintained. For that reason, equation (2.8C) is adjusted to make the $\sigma_{R,allow}$ dependent on the combination of rupture time and temperature, using the Larson-Miller Parameter (*LMP*) substitution from equation (2.8B):

$$\sigma_{R,allow} = 8.2081 \cdot 10^5 \cdot \exp(-3.7920 \cdot 10^{-4} LMP) \qquad , \qquad (2.8D)$$

which is used in the remainder of the report for the creep rupture strength determination for lifetimes larger than 100,000 hours.

2.8.1.3 Stress to obtain maximum allowable creep strain

Another limit determined by the ASME II code, section D, is the average stress that creates a maximum allowable creep strain ($\varepsilon_{c,allow}$) of 0.01% accumulated in 1,000 hours. This is expressed by equation (2.3D) from paragraph 2.3:

$$\varepsilon_{c,allow.} = B_c \left(\frac{\sigma_{c,allow}}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \cdot t$$
 , (2.8E)

where ($\sigma_{c,allow}$) is the maximum allowable stress based on the creep strain rate and t is the exposure time.

2.8.1.4 Comparison of adjusted results from the equations and ASME values

The ASME design code considers that in a certain temperature range one damage mechanism is dominant. After evaluation, the maximum allowable stress for Alloy 617 as function of temperature is defined as shown in figure 2.34 and in table 2G. Transition temperatures exist above which one damage mechanism is dominant and below which another. These transition temperatures and the most dominant damage mechanism in that temperature range are displayed in table 2G as well. Yielding (i.e. plastic deformation caused by a load) is the dominant damage process below 640 °C, while above that temperature creep rupture is the dominant damage process. For the use of Alloy 617 in ASME boiler components, the temperature is maximised at 899 °C by the ASME II material standards code, section D.



Figure 2.34: Comparison of the ASME II, Section II-D, values and calculated values after adjustment.

Temperature range		max $\sigma_{allow.}$ at the	Damage mechanism
Start End		end of temp. range	
[°C]	[°C]	[MPa]	
20	256	160.0	Plasticity ($\frac{2}{3}\sigma_{YS}$)
256	640	142.8	Plasticity ($\sigma_{YS,allow.}$)
640	899	12.3	Creep rupture stress ($\sigma_{R,allow}$)

Table 2G: Dominant damage mechanisms per temperature range.

2.8.2 Fatigue design curve and the application of conservative factors

The fatigue design curve for cyclic processes at temperatures up to 425 °C does not require any additional restrictions or limits to be applied, since the conservative factors as described in paragraph 2.5.1 are already incorporated into this fatigue design curve.

The fatigue curves for cycling events in the temperature range of 538 °C and higher, as described in paragraph 2.5.3, require adjustments based on the restrictions from the ASME II code, Section D [13]. On the value for the number of cycles to failure (N_f), the conservative factors as described in paragraph 2.5.1 have to be applied to account for scatter in data (factor 2.0), differences in grain size (factor 2.5) and environmental conditions (factor 4.0). Therefore, the N_f determined from the best-fit curve has to be divided by 20 to obtain the number of cycles to failure used for the fatigue design curve based on the cycles to failure conservative factor ($N_{f,ASME,N}$):

$$N_{f,ASME,N} = \frac{N_f}{2.0 \cdot 2.5 \cdot 4.0} = \frac{N_f}{20}$$
(2.8F)

Also, the fatigue design curve based on the 'scatter in data' factor on the stress amplitude has to be made. This is done by taking for that stress amplitude the value of the number of cycles to failure at 2.0 times that stress amplitude $N_f(2\Delta\sigma)$, resulting in the fatigue design curve based on the stress amplitude ($N_{f,ASME,\sigma}$):

$$N_{f,ASME,\sigma} = N_f(2\Delta\sigma) \qquad (2.8G)$$

The final fatigue design curve is determined by the lesser value for the number of cycles to failure from the two fatigue design curves as described by equations (2.8F) and (2.8G) making the value for the number of cycles to failure of the final fatigue design curve ($N_{f,ASME}$):

$$N_{f,ASME} = \min(N_{f,ASME,N}, N_{f,ASME,\sigma})$$
(2.8H)

The final fatigue design curve is used for the design calculations.

2.8.3 ASME restrictions on the accumulated creep-fatigue

The design code does not deal with the interaction between creep and fatigue qualitatively, though it recognises the significance of this interaction. All ASME design codes use the diagram as shown in figure 2.35 for the creep-fatigue interaction. The total fatigue damage parameter (D_f) is shown on the horizontal axis in figure 2.35 and the total creep damage parameter (D_c) is displayed on the vertical axis. The total damage accumulation caused by fatigue creates a coordinate on the horizontal axis. The total creep damage parameter results in a coordinate along the vertical axis. Both coordinates form a point in this diagram representing the total damage accumulation point. The ASME design code requires this total damage accumulation point to be positioned underneath the lines as shown in the diagram. The lines of this diagram depend on the intersection point (D_{fm} , D_{cm}). In the ASME code, Alloy 617 is suggested to be treated similar as Ni-Fe-Cr Alloy 800H, which has an intersection point at (0.1,0.1).



Figure 2.35: Creep-fatigue diagram from the ASME code [16].

3. Toolbox

Prediction of the dominant damage mechanism at certain conditions, its magnitude and its effect on the integrity of a component, is one of the key elements for the design of that component. Therefore, models have been created using Microsoft Excel to perform this prediction. These models aim to provide an easy-to-use 'Toolbox' that creates a clear overview of the damage mechanisms and their influence (i.e. dominance) by adjusting the input parameters. The input parameters are applied stress, temperature, exposure time and/or number of expected cycles. Also, the Toolbox may be a helpful aid for life-time estimation.

3.1 Maximum allowable stress

The first check in the Toolbox is whether the maximum allowable stress is exceeded by the sum of the primary stresses. Primary stresses are induced by the process (e.g. pressure) or by the construction (e.g. geometry and weight) and they are commonly constant over time. Secondary stresses are reaction stresses inside the material that relax (i.e. diminish) over time and because of that reducing influence, they are not considered for the maximum allowable stress determination. Other stresses are covered in the Toolbox differently, like (cyclic) thermal stress is limited by the ratchetting criteria.

3.1.1 Toolbox part for maximum allowable stress

For a certain temperature, the value for primary stress is entered into the Toolbox. The Toolbox calculates the maximum allowable stress for that given temperature and directly gives a result if the entered primary stress may continue to the next phase of the assessment or that it is unacceptably high so design action has to be taken to reduce this stress. Five temperatures/stress criteria can be assessed simultaneously in the Toolbox. As is described in paragraph 2.8, one damage mechanism criterion prevails over the others in certain temperature ranges:

Min. Temp.	Max. Temp.	Equation for σ_{allow} as function of temperature				
20 °C	256 °C	⅔ x 240 = 160 (3.1	A)			
256 °C	640 °C	$0.9 * 240(0.6512 + 0.3910 \exp(-5.7516 \cdot (3.1)))$	10 ⁻³ T)) B)			
640 °C 899 °C		$8.2081 \cdot 10^5 \cdot \exp(-9.4801 \cdot 10^{-3}T_K) $ (3.1)	for t ≤ 100,000 hours C1)			
		$8.2081 \cdot 10^5 \cdot \exp(-3.7920 \cdot 10^{-4}LMP) $ (3.1)	for t > 100,000 hours C2)			

in which σ_{allow} is the maximum allowable stress in MPa, T is the temperature in degrees Celsius, T_K is the temperature in Kelvin and *LMP* is the Larson-Miller Parameter.

3.1.2 Limit for cyclic thermal stress and application in the Toolbox

Cyclic thermal stresses are limited to avoid ratchetting and plastic shakedown. The thermal stress range limit (ΔQ_{max}) is calculated via equations (2.6G) and (2.6H), which are repeated here:

$$\Delta Q_{max} = 2\sigma_{y,Tmax} , \quad \text{for } 0 < X < 0.5 \quad , \quad (3.1D) \text{ and}$$

$$\Delta Q_{max} = 4(\sigma_{y,Tmax} - \sigma_p) \quad , \quad \text{for } 0.5 \le X \le 1.0 \quad , \quad (3.1E)$$

With $\sigma_{y,Tmax}$ is the yield strength at the highest temperature present and σ_p is the primary stress acting on the component. These equations have been incorporated into the Toolbox as well.

3.1.3 Degree of constraint

For the development of thermal stress, full mechanical constraint of the boiler component would be a worst-case scenario, since thermal stresses may become that high that the integrity of the component is affected due to severe plastic deformation. This may be due to the design (e.g. in the construction) or by self-constraint caused by a thermal gradient through the wall of the component. Therefore, it may be useful to calculate the maximum allowable degree of constraint in order to prevent the occurrence of unacceptably large plastic deformation. This is especially valid for thicker walled components that may exhibit a thermal gradient across the wall thickness that causes a significant thermal stress. The degree of constraint is given as a percentage that is calculated by dividing the thermal stress range limit (ΔQ_{max}), as determined via equations (3.1D) and (3.1E), by the full constraint stress ($\Delta \sigma_{constraint}$):

% of constraint allowed =
$$\left|\frac{\Delta Q_{max}}{\Delta \sigma_{constraint}}\right| \cdot 100\%$$
 , (3.1F)

in which $\Delta \sigma_{constraint}$ depends on the average elastic modulus (E_{ym}), which is ½x the sum of the elastic modulus at the maximum temperature (T_2) and that at the minimum temperature (T_1), and α_m , which is the coefficient of thermal expansion taken as the average value of the coefficients of thermal expansion at T_1 and T_2 , and the temperature difference between T_1 and T_2 :

$$\Delta \sigma_{constraint} = E_{vm} \cdot \alpha_m \cdot (T_2 - T_1) \tag{3.1G}$$

3.2 Creep prediction

Creep depends on temperature, time and stress magnitude. In paragraph 2.3 and 2.4 the material creep damage for Alloy 617 has been determined as function of these parameters. The creep damage accumulation is used to determine the creep rupture time or strength and to create an expression for the creep strain rate (i.e. the required time component).

3.2.1 Creep time to rupture determination

The creep rupture strength is required for the determination of the maximum allowable stress at temperatures above 640 °C. The quantification of the creep damage is obtained through the ratio of exposure time to the time to rupture. This is done by combining equation (3.1C2) in paragraph 3.1.1 with the Larson-Miller Parameter (*LMP*) from equation (2.4B):

$$LMP = T_K(\log t_R + 20) \qquad (3.2A)$$

into an equation for the time to rupture (t_R) which depends on stress (σ) and temperature (T_K) in Kelvin:

$$t_{R} = 10^{\left(\frac{LMP}{T_{K}} - 20\right)} = 10^{\left(\frac{\ln\left(\frac{\sigma}{8.2081 \cdot 10^{5}}\right)}{-3.7920 \cdot 10^{-4} \cdot T_{K}} - 20\right)} = \frac{\left(\frac{\sigma}{8.2081 \cdot 10^{5}}\right)^{\frac{-6.0723 \cdot 10^{3}}{T_{K}}}}{10^{20}} \quad .$$
(3.2B)

The mathematics for obtaining equation (3.2C) is described in appendix D, paragraph D5.

The creep damage parameter (D_c) is the ratio of the exposure time (t) to the time to rupture, which has to be smaller than 1.0 in order to avoid failure due to creep rupture:

$$D_c = \frac{t}{t_R} < 1.0$$
 (3.2C)

3.2.2 Creep strain caused by primary stress

Primary stresses are assumed constant over time as described in paragraph 3.1. Using equation (2.3D) and the temperature-dependent material parameters B_c , n, μ and the constant stress, the creep strain rate remains unchanged over time and hence the creep strain caused by the primary stress ($\varepsilon_{c,p}$) accumulates linear over time (t):

$$\varepsilon_{c,p} = B_c \left(\frac{\sigma}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \cdot t$$
 (3.2D)

3.2.3 Creep strain influenced by stress relaxation

Due to creep, part of the constant total strain (ε_{total}) converts into creep strain over time ($\varepsilon_{c,s}(t)$) which relaxes the stress over time ($\sigma(t)$) [5,6,64-66,78,79,85,86]:

$$\sigma(t) = E \cdot \left(\varepsilon_{total} - \varepsilon_{c,s}(t)\right) \tag{3.2E}$$

The reduced stress over time ($\sigma(t)$) causes the creep strain rate ($\dot{\varepsilon}_{c,s}$) to diminish over time as well, reducing the influence of the creep mechanism over time (slight adjustment of equation 2.3D):

$$\dot{\varepsilon}_{c,s} = B_c \left(\frac{\sigma(t)}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right)$$
(3.2F)

For the evaluation of creep in an ASME design, time increments may be taken to express the development of creep (see also appendix C of this report). Taking each time increment sufficiently small, the creep strain rate at the beginning of each time increment can be determined rather precisely. The relaxed stress and accumulated creep strain at the end of that time increment are calculated using this creep strain rate and these values are then used to calculate the creep strain rate at the beginning of the subsequent time increment. The creep strain per time increment ($\Delta \varepsilon_{c,s}$) that follows from integration is:

$$\Delta \varepsilon_{c,s}(t) = B_c \cdot \left(\frac{\sigma(t)}{\mu}\right)^n \cdot \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \cdot \Delta t \qquad (3.2G)$$

The creep strain rate declines most rapidly at the beginning of the life-time [6,64-66,78,79,85,86]. Therefore, the time increments at the start are chosen small with the first one typically smaller than one second. Because the creep influence becomes smaller over time due to stress relaxation (i.e. creep strain rate declines), each subsequent time increment may be chosen larger to avoid extensive calculations but it cannot be too large to ensure that the model remains representative for the accumulation of creep strain and the development of creep strain rate. A multiplier of 1.05 has been chosen to achieve that goal (for validation of this multiplier, see paragraph 5.4 of this report). So, the second time increment is 1.05 times larger than the first, the third time increment is in turn 1.05 times larger than the second, etc. The ASME criterion is a maximum creep strain of 0.01% per 1,000 hours and the creep strain caused by the secondary stress ($\varepsilon_{c,s}$) over the first 1,000 hours is the sum of the strain increments over those 1,000 hours:

$$\varepsilon_{c,s} = \sum_{1}^{n_t} \Delta \varepsilon_{c,s}(t) \qquad , \qquad (3.2H)$$

in which n_t is the number of time increment steps, which is determined at:

$$n_t = \frac{\log 3,600,000}{\log 1.05} = 309.4 \xrightarrow{rounded to} 360$$
 , (3.21)

in which 3,600,000 is the number of seconds in 1,000 hours. The first time-increment (n_1) is then:

$$n_1 = \frac{3,600,000s}{1.05^{360}} = 8.48 \cdot 10^{-2}s \tag{3.2J}$$

This is sufficiently small to determine the development of stress relaxation and creep strain accurately (see also the validation in chapter 5). Besides the creep strain over the first 1,000 hours, the Toolbox calculates the total accumulated creep strain. Therefore, the Toolbox calculates the first time increment based on the hold-time, though in order to regard the possible life-span of 30 years, the number of time increment steps is taken as 480 instead of 360.

3.2.4 Maximum allowable creep strain

.

Using equations 3.2D and 3.2H, the combined creep strain from both primary and secondary stress based on temperature is calculated. As described in paragraph 2.8, the maximum allowable creep strain ($\varepsilon_{c,allow}$) is 0.01% per 1,000h (i.e. 10⁻⁴ per 1,000h). Therefore, over the first 1,000 hours the combination of the constant primary stress and the time-dependent diminishing secondary stress has to be such that the sum of the creep strains is:

 $\varepsilon_{c,p} + \varepsilon_{c,s} \leq \varepsilon_{c,allow.}$

(3.2K)

3.3 Fatigue prediction

The fatigue damage prediction starts with identification of the cycle types and cycling events occurring during service. Cycle types include continuous cycling and cycling with a hold-time in tension direction. Common cycling events are cold start, warm start, hot start and load changes during service. Relevant parameters include operating temperatures, pressures, supplemental loads, and expected numbers of cycles. Five different types of cycles and/or cycle events can be inserted simultaneously into the Toolbox and description fields are created in which a clear description of the cycle type or cycle event can be given. The cycle types and cycling events are arranged in stress range size, starting with the largest and ending with the smallest (i.e. max – min cycle counting method). The largest stress ranges and/or amplitudes are expected during the largest temperature difference in time (e.g. during start-up of the HRSG), which may be implemented in the Toolbox by applying the cycle type/cycling event with the largest temperature difference first.

3.3.1 Stress amplitude determination

The stress amplitude is determined using finite element (FE) calculation software. The stress amplitude is one of the input parameters used in the Toolbox that calculates the number of cycles to failure for that stress amplitude based on the fatigue design curve belonging to the temperature of evaluation.

3.3.2 Fatigue for temperatures below 425 °C

As described in paragraph 2.5.2 of this report, a standard fatigue design curve for Alloy 617 for temperatures below 425 °C exists. This curve may be used directly. To obtain the number of cycles to failure (N_f) from the stress amplitude applied ($\Delta \sigma$), equation (2.5E) has been used:

$$N_f = \left(\frac{E \cdot \log((A_f))}{4(\Delta \sigma - B_f)}\right)^2 \tag{3.3A}$$

This equation cannot be used directly due to the asymptotic behaviour of this equation with stresses around the endurance limit (B_f). This behaviour has been diverted by using multiple equations, which is equation (3.3A) for the low-cycle fatigue part, best-fit plots for the part around the endurance limit, and a linear relation for the high-cycle fatigue part (i.e. no plastic deformations, only slight cyclic hardening). These equations have been incorporated into the Toolbox and the Toolbox selects the correct equation based on the stress amplitude.

3.3.3 Fatigue for temperatures above 538 °C

Fatigue curves are constructed for three temperature ranges, i.e. 538-704 °C, 704-871 °C and 871-982 °C, which are described by the equations in paragraph 2.5.3. The Toolbox considers the conservative factors enforced upon by the design code, as described in paragraph 2.8.2, and converts these fatigue curves to fatigue design curves. The Toolbox calculates the number of cycles to failure of the fatigue design curve directly ($N_{f,ASME}$).

3.3.4 Fatigue damage

The Toolbox determines the number of cycles to failure for the kth cycle conditions ($N_{f,k}$) based on the stress amplitude and the fatigue design curve at that temperature. This value is compared to the number of cycles that is expected for the kth cycle conditions (n_k) and the ratio between these two values is the fatigue damage parameter induced by those specific conditions ($D_{f,k}$):

$$D_{f,k} = \frac{n_k}{N_{f,k}} \tag{3.3B}$$

This can be done for five separate stress amplitudes and temperature combinations (besides 'k', also 'l', 'm', 'n' and 'p') and their sum is the total damage parameter caused by fatigue. The value for the total damage fatigue parameter ($D_{f,total}$) must be smaller than 1.0 to ensure that the material is able to withstand the cycling conditions:

 $D_{f,total} < 1.0$. (3.3C)

3.4 Creep-fatigue determination

Creep-fatigue is the interaction between the two high-temperature damage mechanisms creep and fatigue. It is hard to quantify the synergy effect of these damage mechanisms, though a correlation between the separately determined creep damage parameter and fatigue damage parameter may be used. The determination of the creep damage parameter is expressed in paragraph 3.2.1, equation (3.2C) and the fatigue damage parameter is calculated by equations (3.3B) and (3.3C) in paragraph 3.3.4.

3.4.1 ASME requirement fulfilment

The ASME design code has provided the creep-fatigue diagram (see figure 2.35 in paragraph 2.8.3) in which the creep damage parameter (D_c) and the fatigue damage parameter (D_f) are the coordinates for the damage accumulation point. The ASME design code requires that the coordinates of this point are underneath the two linear lines that are plotted through the coordinates (0.1,0.1) and (0,1) and through the coordinates (0.1,0.1) and (1,0). These lines can be described and since the graph has the fatigue damage parameter along the horizontal axis it is convenient to express the maximum allowed creep damage parameter ($D_{c,max}$) as a function of the fatigue damage parameter. The line trough (0.1,0.1) and (0,1) is written as:

$$D_{c,max} = -9D_f + 1$$
 , (3.4A)

which is used for $D_f \le 0.1$. The line through (0.1,0.1) and (1,0) is used for $D_f > 0.1$ and can be expressed:

$$D_{c,max} = -\frac{D_f}{9} + \frac{1}{9}$$
(3.4B)

The ASME requirement is fulfilled in case D_c is smaller than or equal to $D_{c,max}$:

$$D_c \le D_{c,max} \tag{3.4C}$$

Equations 3.4A and 3.4B can also be written with the creep damage parameter and fatigue damage parameter interchanged, though then $D_{f,max}$ is calculated from D_c .

3.5 Toolbox output

The input parameters primary and secondary stress, temperature, exposure time and number of cycles are entered in the orange coloured fields on the first worksheet of the Toolbox. The Toolbox calculates the material properties based on these input values using the equations and plots gathered in worksheets beyond the first worksheet. The calculated material property values are returned to the first worksheet in the light-green coloured fields. In the blue coloured fields, the Toolbox gives the maximum allowable values as calculated based on the regulations and limits set by the design code. The maximum allowable values are compared with the input values and with the calculated results. Based on these comparisons, the Toolbox gives the result "TRUE" in a dark-green field, which indicates that the input value or calculated result is lower than the maximum allowable value. In case the input value or calculated result exceeds the maximum allowable value, the result is a "FALSE" in a red field. With the result "TRUE" on all six comparisons, the material and design conditions are met and the design should be acceptable for ASME certified use.

3.5.1 Comparison 1: primary stress versus the maximum allowable stress

The maximum allowable stress depends on the dominant damage mechanism at that temperature. This is determined and calculated by the Toolbox and the result is compared with the primary stress input parameter value. In case the input value is lower than the maximum allowable stress value, the Toolbox returns the result "True" and the next comparison can be investigated. When the input value exceeds the maximum allowable stress value, the Toolbox gives the result "FALSE" and a primary stress value lower than the maximum allowable stress has to be entered. Since the primary stress is a result from FE calculations, the design or service parameters must be adjusted to lower the primary stress. Examples of adjustments are changing the wall thickness, different geometry, reduction of internal pressure and lowering the service temperature.

3.5.2 Comparison 2: maximum allowable thermal stress to avoid ratchetting

The avoidance of ratchetting determines the maximum value for the thermal stress. The thermal stress limit is calculated by the Toolbox via equations 3.1D and 3.1E and the calculated value is compared with the input value for the thermal stress. Just as described at comparison 1 in the previous sub-paragraph, a "TRUE" means that the input value is allowed to be used and a "FALSE" requires a lower input value for the thermal stress. This may be achieved by lowering the primary stress (see comparison 1), or by reducing the degree of constraint, improving heat distribution or slower heating.

3.5.3 Comparison 3: time to rupture is longer than the exposure time

Based on the input parameters stress and temperature, the time to rupture is calculated and displayed on the first worksheet. In case the input parameter exposure time is smaller than the time to rupture, the Toolbox provides a "True". With a "False", an input parameter must be lowered. On the first worksheet, also the allowable creep rupture strength is provided for the entered time – temperature combination, which gives an indication for the stress value required to withstand creep rupture.

The creep damage parameter is also determined. This is done by dividing the value of the exposure time by the value of the time to rupture. The creep damage parameter is determined for five creep events (i.e. combination of temperature, stress and time) simultaneously and the sum of these damage parameters is considered in comparison 6 for the creep-fatigue interaction.

3.5.4 Comparison 4: exceeding of the maximum allowable creep strain

The creep strain caused by primary stress (thus constant over time) and secondary stress (which relaxes over time) are calculated based on these stresses, the exposure time and the temperature. The total creep strain developed during the first 1,000 hours is compared with the maximum allowable creep strain per 1,000 hours. The result "TRUE" grants passage to the next comparison, the result "FALSE" requires a lower input value for one of the input parameters. In comparison 1 and 2 it is described how to reduce the respective primary and secondary stress. Lowering the temperature is also a possibility to limit the creep strain.

The values for the creep strain developed during the exposure time and the ratio between the creep strain caused by primary stress to that caused by secondary stress, are returned to the first worksheet for reference.

3.5.5 Comparison 5: maximum allowable fatigue damage accumulation

The fatigue damage is calculated on a separate worksheet, because the fatigue cycles have to be described carefully. Five separate cycle situations can be described and quantified simultaneously. Data are gathered from the service process (e.g. number of cycles, pressure fluctuations/differences and temperature ranges) and stresses amplitudes are calculated by the FE software program. For each of the five cycle situations, the input value for the number of cycles is compared with the fatigue design curve related to that cycle situation. The fatigue damage parameter for each cycle situation is calculated by dividing the number of cycles input value by the number of cycles to failure as determined from the fatigue design curve. Then the fatigue damage parameters of each cycle situation are summed to obtain the total fatigue damage parameter which has to be lower than 1.0 and is used in the next comparison. The first worksheet illustrates this with a "TRUE" result. In case the total fatigue damage parameter is higher than 1.0, the result is "FALSE" and the input values have to be adjusted. Options to reduce the fatigue damage are lowering the stress amplitude by design or diminishing the number of cycles occurring during service.

3.5.6 Comparison 6: fulfilment of ASME requirement for creep-fatigue

The total damage accumulation parameters as determined for creep in comparison 3 and for fatigue in comparison 5 are used by the Toolbox as coordinates to create the damage accumulation point in the ASME creep-fatigue diagram as illustrated in figure 2.35 in paragraph 2.8.3. This point has to be situated underneath the two linear lines as equated with equations (3.4A) and (3.4B). The result "TRUE" is returned in case the damage accumulation point has coordinates underneath these two linear lines. When "FALSE" is given, one or both total damage accumulation parameter(s) has/have to be lowered. Lowering these parameters is similar to that for creep in paragraph 3.5.3 and for fatigue in paragraph 3.5.5.

3.6 Functionality of the Toolbox

The Toolbox is able to quantify damage accumulation due to creep and fatigue at high temperatures and to indicate the onset of plastic deformation at lower temperatures. Also the comparison with the limits/requirements as set by the design code is adequate and valid. The most interesting observations about Alloy 617 and limitations of the Toolbox are summarised in this paragraph.

3.6.1 Observations about Alloy 617

Alloy 617 is an important candidate material for service at high temperatures up to ca. 900 °C. The FCC microstructure makes that Alloy 617 has a good strain-hardenability and is therefore ductile with elongation until failure percentages above 60 at room temperature and well over 40 at around 900 °C [29]. Due to the good strain-hardenability, the ultimate tensile strength of Alloy 617 is at least a factor 2 higher than the yield strength regardless the temperature. For that reason, the ultimate tensile strength of Alloy 617 is not a controlling factor for the maximum allowable stress determination, since the maximum allowable stress values based on the yield strength are lower in the entire temperature range. The yielding damage mechanism is therefore the dominant damage mechanism in the lower temperature range.

Nickel-based alloys, like Alloy 617, have a high melting temperature which in turn leads to a high homologous temperature above which creep processes become significant. Also, the high activation energy for diffusion in nickel-based alloys makes diffusional creep more difficult. The solute atoms and precipitates in Alloy 617 are hindering the dislocation movement, making it even more creep resistant. This good creep resistance is shown in the Toolbox by the high creep rupture strength values and by the relatively low creep strain rates in the high temperature region.

The creep strain allowed to accumulate in a certain time-period is limited because else the material would not have time to respond to the fast damage accumulation in case creep develops rapidly. For example, wedges can develop at the grain boundaries, since the dislocation lines do not have time to glide towards another slip plane upon encountering a dislocation obstacle. This damage mechanism becomes dominant at temperatures over 800 °C.

3.6.2 Limitations of Toolbox

Some limitations have been observed when using the Toolbox.

Fatigue curves for Alloy 617 have not been determined for temperatures above 982 °C. Also, the design code for main boiler components, the material standards code from the ASME, does not allow Alloy 617 to be used beyond 899 °C. For that reason, 899 °C is the maximum temperature for which this Toolbox may be used.

Another drawback is that fatigue data is not available for the temperature range 425 – 538 °C. This makes the determination of fatigue damage in this temperature range not possible. It may be considered to use a 'worst-case' scenario by applying the 538 °C fatigue design curve instead.

4. ANSYS simulations

A material model has been constructed in ANSYS based on the material data as described in chapter 2 of this report. The most suitable material model for describing Alloy 617 in the temperature range from room temperature up to 800 °C is a model that is able to deal with damage mechanisms like plasticity, creep and fatigue in a quantitative manner. Alloy 617 is a nickel-based alloy that is both precipitate- and solution-hardened requiring hardening effects to be taken into account. For the true stress – true strain relation the Materials Properties Council (MPC) model is used, which distinguishes between micro-region deformation and macro-region deformation and which is also suggested by the ASME design code.

Also, a suitable test specimen set-up has been created in ANSYS which reflects the set-up as would be used in an actual test. The results of the ANSYS simulations on the test specimen using the material model are used to validate the results from the Toolbox calculations, which is done in chapter 5.

4.1 Test specimen for FE simulations

The standard round tensile test specimen for uniaxial testing from the ASTM A370/ASTM E8 [89,90] has been chosen for the ANSYS simulations. All materials listed in the ASME II code are tensile tested per this method. Also, most data taken from the literature have been determined using this method. Therefore, it is a suitable configuration for ANSYS simulations.

The area of interest in a standard round tensile specimen is shown in figure 4.1. The specimen has been designed in this manner to ensure thinning, necking and failure to occur in the region indicated by 'G', and hence is measured by the gauge (i.e. tool for measuring the elongation, which is often used in a tensile test).



				DIMENSION	VS					
	Standard	Specimen		Small-Size Specimens Proportional to Standard						
Nominal Diameter	in.	mm	in.	mm	in.	mm	in.	mm	in.	mm
	0.500	12.5	0.350	8.75	0.250	6.25	0.160	4.00	0.113	2.50
G-Gauge length	2.00±	50.0 ±	1.400±	35.0 ±	1.000±	25.0 ±	0.640±	16.0 ±	0.450±	10.0 ±
	0.005	0.10	0.005	0.10	0.005	0.10	0.005	0.10	0.005	0.10
D-Diameter (Note 1)	0.500±	12.5±	0.350±	8.75 ±	0.250±	6.25 ±	0.160±	4.00 ±	0.113±	2.50 ±
	0.010	0.25	0.007	0.18	0.005	0.12	0.003	0.08	0.002	0.05
R-Radius of fillet, min	3/8	10	1/4	6	3/16	5	5/32	4	3/32	2
A—Length of reduced section, min (Note 2)	21/4	60	13/4	45	11/4	32	3/4	20	5/8	16

Figure 4.1: Standard 12.5 mm round tension test specimen with 50mm gauge length [89].

Tensile test machines often have differences in clamping systems. For that reason, five different setups for specimen ends have been developed by the ASTM A370/ASTM E8 as is shown in figure 4.2.



DIMENSIONS

	Specimen 1		Specimen 2		Specimen 3		Specimen 4		Specimen 5	
	in.	mm	in.	mm	in.	mm	in.	mm	in.	mm
G-Gauge length	2.000± 0.005	50.0 ± 0.10	2.000± 0.005	50.0 ± 0.10	2.000± 0.005	50.0 ± 0.10	2.000± 0.005	50.0 ± 0.10	2.00± 0.005	50.0 ± 0.10
D-Diameter (Note 1)	0.500 ± 0.010	12.5± 0.25	0.500 ± 0.010	12.5± 0.25	0.500 ± 0.010	12.5± 0.25	0.500 ± 0.010	12.5± 0.25	0.500± 0.010	12.5 ± 0.25
R-Radius of fillet, min	3/8	10	3/8	10	1/16	2	3/8	10	3/8	10
A—Length of reduced section	2¼ , min	60, min	21/4 , min	60, min	4, ap- proxi- mately	100, ap- proxi- mately	21/4 , min	60, min	21/4 , min	60, min
L-Overall length, approximate	5	125	51/2	140	51/2	140	43/4	120	91/2	240
B—Grip section (Note 2)	1%, ap- proxi- mately	35, ap- proxi- mately	1, ap- proxi- mately	25, ap- proxi- mately	3/4 , ap- proxi- mately	20, ap- proxi- mately	1/2 , ap- proxi- mately	13, ap- proxi- mately	3, min	75, min
C-Diameter of end section	3/4	20	3/4	20	23/32	18	7/8	22	3/4	20
E-Length of shoulder and fillet section, approximate			5/8	16	*(*(*)	• • •	3/4	20	5/8	16
F-Diameter of shoulder		22.2	5/8	16	20232	102.02	5/8	16	19/32	15

Figure 4.2: Types of ends for standard round tension test specimens [89].

From these set-ups, number 3 is selected since this is one of the most frequently used set-ups and it requires less drawing efforts and calculations by ANSYS. This set-up is created in ANSYS and to show the stress distribution throughout the specimen, a ¼ of the specimen has been simulated as shown in figure 4.3.



Figure 4.3: View of the test specimen used in the ANSYS simulations.

When a uniaxial nominal stress is applied (i.e. the ends of the specimen move apart), the location of the maximum and minimum stresses in the specimen are indicated and the stress distribution is shown by a colour legend. Examples are shown in figure 4.4 and 4.5 for an applied stress of 141.76 MPa (i.e. 0.9x the yield strength at 800 °C, arbitrary chosen).



Figure 4.4: Stress distribution in the test specimen after application of a uniaxial nominal stress.



Figure 4.5: Detail of figure 4.4, seen from cross-sectional side.

The speed of testing shall be such that accurate load and strain readings are possible [89,90]. There are several methods to express the testing speed, though since an actual tensile test machine is not used, only the stress rate or strain rate can be used. It is chosen to use the cross-head displacement method, which means that the heads of the specimen move apart with a fixed strain rate. Up to ½ the specified yield strength, the strain rate can be chosen at convenience [89,90]. Since it is a simulation, this strain rate is taken the same as above ½ the specified yield strength. Above ½ the specified yield strength, the strain rate shall not exceed 1.6 mm (1/16 inch) per minute per 25 mm (1 inch) of reduced section [89]. When choosing the specified minimum length for the reduced section, i.e. 60 mm (2¼ inch), which is dimension 'A' in figures 4.1 and 4.2, the maximum strain rate is determined at 3.6 mm (9/64 inch) per minute. The minimum strain rate for testing is set at 0.1x the maximum strain rate, i.e. 0.36 mm (9/640 inch) per minute [89]. The actual strain rate for testing is selected between these minimum and maximum values.

For the selection of the strain rate for the cross-head displacement testing it is chosen to use the maximum strain rate (i.e. 3.6 mm per minute over 60 mm reduced section). This results in a test strain rate ($\dot{\epsilon}_{test}$) of:

$$\dot{\epsilon}_{test} = \frac{3.6mm \cdot min^{-1}}{6 \cdot 10^{1} mm} \cdot \frac{1min}{60s} = 1 \cdot 10^{-3} s^{-1} \qquad (4.1A)$$

This value is also the test strain rate that is often observed in the literature for the performed tests.

4.2 Additional physical material properties required for FE calculations

ANSYS requires additional material properties equated as a function of the temperature to perform calculations and run simulations. First, the coefficient of thermal expansion is determined, which influences the thermal stress. The coefficient of thermal expansion (α) is [29,88]:

$\alpha = -9.46 \cdot 10^{-6}T^2 + 1.13 \cdot 10^{-2}T + 1.06 \cdot 10^{1}$	for T ≤ 600 °C,	(4.2A)
$\alpha = -1.00 \cdot 10^{-5} T^2 + 2.10 \cdot 10^{-2} T + 5.00 \cdot 10^0$	for T ≥ 600 °C,	(4.2B)

in which T is the temperature in degrees Celsius and α has the unit [μ m^{-1.o}C⁻¹].

Second, the Poission's ratio (v) is required for ANSYS, since it is used to determine the amount of elastic contraction in the direction(s) perpendicular to the direction that is being elastically and uniaxially elongated or vice versa. For Alloy 617 the Poisson's ratio is 0.31 in the entire temperature range [29,88].

The third required material property is the thermal conductivity since this determines the thermal gradient over any wall thickness and therewith the resistance to the heat being transferred from the outer surface of the pressurised component to the inner surface or vice versa. This thermal gradient causes (additional) stresses in the material that ANSYS has to account for. The thermal conductivity (k_c) is given in $[W \cdot m^{-1.0}C^{-1}]$ and is equated as function of temperature *T* in degrees Celsius [29,88]:

$$k_c = 1.55 \cdot 10^{-2}T + 1.31 \cdot 10^1 \qquad (4.2C)$$
The density (ρ) of Alloy 617 is 8360 kg·m⁻³ [29,88], which is the fourth additional required property.

The fifth and last required material property is the specific heat (C_p) in [J·kg⁻¹·K⁻¹] or [J·kg⁻¹·OC⁻¹] [88]:

$$C_p = 2.56 \cdot 10^{-1}T + 4.13 \cdot 10^2$$
 , (4.2D)

with temperature T in degrees Celsius.

4.3 Materials Properties Council Model

The Materials Properties Council (MPC) model is used in calculations for constructing stress-strain curves including case strain hardening characteristics. The main equation is [87]:

$$\varepsilon_{ts} = \frac{\sigma_t}{E} + \gamma_1 + \gamma_2 \qquad , \tag{4.3A}$$

in which ε_{ts} is the true strain, σ_t is the true stress, *E* is the Young's modulus at the temperature of interest, γ_1 and γ_2 are the true strains in the respective micro-strain region (i.e. plastic strain at interatomic level) and macro-strain region (i.e. plastic straining on macroscopic scale which can be determined with a measuring tool) of the stress-strain curve.

The parameters σ_t , γ_1 and γ_2 are equations by themselves. The expression for σ_t is:

$$\sigma_t = (1 + \varepsilon_{es})\sigma_{es} \qquad , \tag{4.3B}$$

in which ε_{es} and σ_{es} are the engineering strain and engineering stress respectively [87].

The strain in the micro-strain region (γ_1) is equated to [87]:

$$\gamma_1 = \frac{\varepsilon_1}{2} (1.0 - \tanh[H])$$
 , (4.3C)

with ε_1 being the true plastic strain in the micro-strain region of the stress-strain curve:

$$\varepsilon_1 = \left(\frac{\sigma_t}{A_1}\right)^{\frac{1}{m_1}}$$
 , (4.3D)

and H is the Prager-Drucker factor, which is used to establish plastic yielding:

$$H = \frac{2[\sigma_t - (\sigma_y + K\{\sigma_{uts} - \sigma_y\})]}{K(\sigma_{uts} - \sigma_y)}$$
(4.3E)

In equation (4.3D), A_1 is the curve-fitting constant of the stress-strain curve (micro-strain):

$$A_1 = \frac{\sigma_y(1+\varepsilon_y)}{\left(\ln[1+\varepsilon_y]\right)^{m_1}} , \qquad (4.3F)$$

and m_1 is the curve-fitting exponent for the stress-strain curve equal to the true strain at the proportional limit and the strain hardening coefficient in the large strain region:

$$m_1 = \frac{\ln[R] + (\varepsilon_p - \varepsilon_y)}{\ln\left[\frac{\ln[1 + \varepsilon_p]}{\ln[1 + \varepsilon_y]}\right]} \qquad (4.3G)$$

In equations (4.3F) and (4.3G), σ_y is the engineering yield strength at the temperature of the assessment, ε_y is the 0.2% engineering offset strain (so value is 2.0·10⁻³), R_σ is the ratio σ_y to engineering tensile stress (σ_{uts}) at assessment temperature:

$$R_{\sigma} = \frac{\sigma_y}{\sigma_{uts}} \qquad , \tag{4.3H}$$

and ε_p is the engineering offset strain for the proportional limit, which has a value of 2.0 10⁻⁵ for superalloys, like Alloy 617 [87].

Most of these parameters are also used for the Prager-Drucker factor (H) in equation (4.3E), though this equation also includes model parameter K, which gives an indication for the strain-hardenability:

$$K = 1.5R_{\sigma}^{1.5} - 0.5R_{\sigma}^{2.5} - R_{\sigma}^{3.5} \qquad (4.31)$$

The strain in the macro-strain region (γ_2) is calculated by [87]:

$$\gamma_2 = \frac{\varepsilon_2}{2} (1.0 + \tanh[H])$$
 , (4.3J)

with ε_2 is the true plastic strain in the macro-strain region of the stress-strain curve:

$$\varepsilon_2 = \left(\frac{\sigma_t}{A_2}\right)^{\frac{1}{m_2}}$$
 , (4.3K)

and H is the Prager-Drucker factor as shown in equation (4.3E).

In equation (4.3K), A_2 is the curve-fitting constant of the stress-strain curve (macro-strain):

$$A_2 = \frac{\sigma_{uts} \exp[m_2]}{m_2^{m_2}} ,$$
 (4.3L)

and m_2 is the curve-fitting exponent for the stress-strain curve equal to the true strain at the true ultimate tensile strength, which for superalloys, like Alloy 617 is:

$$m_2 = 1.90(0.93 - R_\sigma) . \tag{4.3M}$$

Using these equations, the finite element (FE) program ANSYS is able to construct isothermal true stress – true strain curves. These curves are used by ANSYS to calculate stress – strain situations.

4.3.1 Check 1: stress exceeding the maximum allowable yield strength

ANSYS calculates a stress that results from either an applied force, thermal expansion combined with mechanical constraint or any other cause. This stress is maximised by the stress allowed by the design code. For a nickel-based superalloy, like Alloy 617, the maximum based on the yield strength is determined by either $\frac{3}{2}$ of the yield strength at room temperature or 90% of the yield strength at temperature (see paragraph 2.8). Therefore, the limit incorporated into ANSYS has the equation:

$$\sigma_{allow,y} = \min\left(\frac{160}{216(0.65123 + 0.39097 \exp(-5.7516 \cdot 10^{-3}T))}\right)$$
[MPa] ,
(4.3N)

where $\sigma_{allow,y}$ is the allowable stress based on the yield strength and T is the temperature in Celsius.

In ANSYS, the option is created to compare the calculated stress resulting from the simulation with the maximum allowable stress. The location of the maximum obtained calculated stress in the object or test sample is indicated by a red dot in the illustration. The ratio between the calculated stress value and the maximum allowable stress value is given by both a number (i.e. normalised value) and a colour indication based on a legend. In this legend, blue is no or hardly any calculated stress and red means that the calculated stress exceeds the maximum allowable stress based on the yield strength. In figure 4.6, an example is shown for the situation at 800 °C with an applied stress equal to the maximum allowable stress based on the yield strength. Note that for this specific example at this temperature the yield strength is not the limiting factor, but the maximum allowable stress based on creep is. Creep is discussed in the next section of this chapter.



Figure 4.6: Normalised stress check for maximum allowable stress based in yield strength at 800 °C for 141.51 MPa (i.e. the minimum yield strength at this temperature).

4.4 Creep mechanism incorporated into the material model

With plasticity covered by the MPC model, the creep mechanism is incorporated into the material model. For Alloy 617, creep damage is determined by the creep rupture strength, which is the stress required to cause failure when the material is exposed to a fixed temperature for a certain amount of time. Additionally, the creep strain developing during any 1,000 hours is not allowed to surpass 0.01%.

4.4.1 Time to rupture

The creep damage equation contains the time to rupture (t_R) , which can be used to show the influence of stress and temperature on the damage allowed to accumulate. The time to rupture can be compared to the exposure time of a component to a certain stress and temperature. The time to rupture is calculated by ANSYS by using equation (3.2B):

$$t_R = \frac{\left(\frac{\sigma}{8.2081 \cdot 10^5}\right)^{\frac{-6.0723 \cdot 10^3}{T_K}}}{10^{20}} \qquad [h] \qquad . \tag{4.4A}$$

An example of an ANSYS calculation for the time to rupture is shown in figure 4.7. For a temperature of 800 °C and an applied constant stress of 26.41 MPa. This stress has been chosen because it is the required creep rupture strength value as calculated by the Toolbox for 30 years at 800 °C. The result shown in this figure is approximately 30 years.



Figure 4.7: Time to rupture in years for 26.41 MPa at 800 °C.

4.4.2 Check 2: time to rupture exceeding the exposure time of the component

One of the input parameters for the model is the exposure time of the component. The stress and temperature combined should create a time to rupture that is longer than that exposure time. An easy and simple overview is created by showing the normalised value in ANSYS, which is done by dividing the time to rupture as determined by ANSYS by the exposure time. In this comparison, the exposure time is shown as a ratio of the calculated time to rupture. In case the normalised value is larger than unity (i.e. "1"), the workpiece in the simulation colours red.

The example shown in figure 4.7 is used to perform the check of the normalised value for the time to rupture and the result is shown in figure 4.8. As expected from the calculated value for the time to rupture in figure 4.7, the normalised value exceeds unity ("1") slightly, which is the ratio between 30 years (i.e. the input parameter exposure time) and 30.02 years (i.e. the calculated result for the time to rupture from ANSYS). The validation of the Toolbox by the material model in ANSYS is described in chapter 5 of this report.



Figure 4.8: Check of normalised value of the exposure time and the time to rupture.

4.4.3 Creep strain

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First, the creep strain rate is calculated using the Norton (i.e. power-law) creep equation (see also equation 2.3D):

$$\dot{\varepsilon}_{c} = B_{c} \left(\frac{\sigma}{\mu}\right)^{n} \exp\left(\frac{-Q_{c}}{R \cdot T_{K}}\right) \qquad , \qquad (4.4F)$$

with $\dot{\varepsilon}_c$ is the steady state creep strain rate, B_c is a temperature-dependent constant, σ is the applied stress, μ is the shear modulus at temperature, and n is a material exponent, Q_c is the activation energy for creep, R is the gas constant, T_K is the absolute temperature (in K). The Norton's creep equation in ANSYS has a similar form and is expressed as:

$$\dot{\varepsilon}_c = C_1 \sigma^{C_2} \exp\left(\frac{-C_3}{T_K}\right) \qquad , \qquad (4.4G)$$

in which C_1 , C_2 and C_3 are the input parameters that ANSYS requires to perform the calculations. The input parameters for ANSYS are equated as a function of the temperature by combining equations (4.4F) and (4.4G):

$$C_{1} = \frac{B_{c}}{\mu^{n}} \qquad [s^{-1}MPa^{-n}] , \qquad (4.4H)$$

$$C_{2} = n \qquad [dimensionless] , \qquad and \qquad (4.4I)$$

$$C_{3} = \frac{Q_{c}}{R} = \frac{410,000}{8.3145} = 49,311.4 \quad [K] \qquad . \qquad (4.4J)$$

ANSYS bases the calculations of C_1 and C_2 on degrees Celsius and those for C_3 on Kelvin. This cannot be altered in ANSYS. Besides the temperature, ANSYS needs the applied stress (σ) as input parameter.

 B_c , μ and n are determined by temperature-dependent equations as shown in paragraph 2.3, and for that reason, these input parameters have to be entered in ANSYS as such. The value of material exponent n is described by two equations, one for temperatures below 800 °C and one for above 800 °C. Since the model is developed for temperatures up to 800 °C, only equation (2.3F1) is used in the material model for n and C_2 , which is repeated here:

 $C_2 = n = 1.3707 \cdot 10^{-4} T^2 - 2.3924 \cdot 10^{-1} T + 1.1058 \cdot 10^2 \quad , \tag{4.4K}$

with T is the temperature in degrees Celsius.

The equations for C_1 and C_2 are only accurate in the time-dependent temperature range (i.e. creep region). Above 530 °C, creep is becoming measurable, though creep is becoming significant above 640 °C (see chapter 2).

The Toolbox distinguishes between constant primary and relaxing secondary stresses since these stresses influence the creep strain rate differently. Both situations occur often simultaneously. The FE program ANSYS is not able to differentiate between primary and secondary stresses, but uses an algorithm to calculate the creep strain rate. The output is the total creep strain, which is caused by the combined primary and secondary stress.

For the validation of the Toolbox, appropriate boundary conditions have been set in ANSYS. First, in figure 4.9, the creep strain is shown that has accumulated after 30 years at 801 °C, exposed to a primary stress of 31.04 MPa (i.e. the maximum stress allowed) only. The maximum creep strain that is allowed to accumulate according to the Toolbox is $0.01\%/1,000h \times 1h/3,600s \times 946,080,000s = 0.02630\%$, which approaches the value as calculated by ANSYS as shown in figure 4.9. For the complete validation, see chapter 5.



Figure 4.9: Accumulated creep strain after 30 years at 801 °C and at a constant stress of 31.04 MPa.

Also, the boundary condition of only a secondary stress has been evaluated for the validation of the Toolbox. An example is shown in figure 4.10, where a constant strain of $5.087 \cdot 10^{-4}$ is applied (i.e. the resulting strain after applying a stress of 80 MPa and then holding the sample in that position) followed by a hold for 100,000 hours at a temperature of 800 °C. The value for the accumulated creep strain is $4.317 \cdot 10^{-4}$ according to the Toolbox, which is almost equal to the value of $4.31097 \cdot 10^{-4}$ as calculated by ANSYS (see figure 4.10).



Figure 4.10: Accumulated creep strain after 100,000 hours at 800 °C at a constant strain of 5.087 10⁻⁴.

4.4.4 Check 3: creep strain exceeding the maximum allowable creep strain

ANSYS calculates the total accumulated creep strain. This value is compared to the creep strain that is allowed to accumulate during that time, which is done by multiplying the maximum allowable creep strain of 10⁻⁷ per hour (i.e. maximum allowable creep strain of 0.01% per 1,000 hours as set by the ASME design code) with the exposure time. In case the exposure time is longer than 1,000 hours, the creep strain developed over the first 1,000 hours is determined, since that is the period in which commonly most of the creep strain develops. The value of the creep strain is often a rather unstraightforward value which does not provide an easy overview. For that reason, it is chosen to obtain a normalised value, which is performed by ANSYS by dividing the calculated creep strain value by the maximum allowable creep strain for that exposure time.

In figure 4.11, the normalised value check is shown of the location, stress, temperature and time as described for figure 4.9. This overview shows that the calculated strain by ANSYS is approximately the same as the maximum allowed value (i.e. deviation is about 0.03%), which is the purpose of this check. The creep strain values as determined by ANSYS are the same values as the values calculated by the Toolbox for both maximum allowable creep strain and actual creep strain accumulated. The validation of the Toolbox is described in chapter 5.



Figure 4.11: Check of the normalised value of the accumulated creep strain.

4.5 Fatigue mechanism incorporated into the material model

The next damage mechanism added to the material model is fatigue. Most convenient is to describe fatigue at lower temperatures (i.e. lower than and equal to 425 °C) by the stress – life cycles method and fatigue at higher temperatures (i.e. higher than and equal to 538 °C) by the strain – life cycles method (see paragraph 2.5). For the temperature range between 425 °C and 538 °C, data has not been generated and hence a fatigue design curve does not exist for this temperature range.

4.5.1 Fatigue at lower temperatures (below 425 °C)

The fatigue design curve determined for lower temperatures is constructed based on data that is determined from fully reversed experiments (i.e. the R-ratio is -1), which means that the mean stress is zero. The design curve constructed in ANSYS is therefore also meant for fully reversed load cycles.

In the Toolbox, 75 stress – cyclic life points have been determined, which have been exported to ANSYS. The fatigue design curve constructed by ANSYS based on these points is identical to that plotted by the Toolbox (see also the validation in the next chapter of this report). With this data, ANSYS is able to determine the number of cycles to failure for any stress amplitude.

4.5.2 Check 4: the number of cycles to failure (below 425 °C)

ANSYS displays the number of number of cycles to failure for any entered stress amplitude. In figure 4.12, number of cycles to failure is shown for a stress amplitude of 245 MPa. The calculated value from the Toolbox for the cycles to failure is $1.49 \cdot 10^5$, which is identical to the value as calculated by ANSYS.



Figure 4.12: Calculated number of cycles to failure for the stress amplitude of 245 MPa.

4.5.3 Fatigue at higher temperatures (above 538 °C)

For the evaluation of fatigue at temperatures above 538 °C, the strain – cycle life assessment method is used. ANSYS requires six parameters to perform this assessment, i.e. the strength coefficient (σ'_f), the strength exponent (b_f), the ductility coefficient (ε'_f), the ductility exponent (c_f), the cyclic strength coefficient (K'), and the cyclic strain hardening exponent (n').

ANSYS requires all six parameters because the values for the cyclic strength coefficient and cyclic strain hardening exponent can be taken directly from literature sources. For the input parameters in ANSYS, the cyclic strength coefficient and cyclic strain hardening exponent are calculated using equations (2.5J) and (2.5K) and entering the values of the other four fatigue parameters. Based on the values of the fatigue parameters as determined in paragraph 2.5.3, table 4A has been constructed.

Т	σ'_f	b _f	ε'_{f}	C _f	K'	n'
[°C]	[Pa]	[-]	[-]	[-]	[Pa]	[-]
538-704	1.330E+09	-0.11	0.68	-0.76	1.406E+09	0.145
704-871	1.160E+09	-0.09	0.68	-0.76	1.214E+09	0.118
871-982	1.085E+09	-0.08	0.68	-0.76	1.130E+09	0.105

Table 4A: Values of the fatigue parameters in different temperature ranges.

ANSYS calculates the strain amplitude (ε_a) based on the stress amplitude ($\Delta \sigma_a$) and accounts for cyclic hardening influences using this relation:

$$\varepsilon_a = \frac{\Delta \sigma_a}{E} + \frac{1}{2} \left(\frac{\Delta \sigma_a}{K'} \right)^{\frac{1}{n'}}$$
(4.5A)

In an integrated black-box, ANSYS constructs the fatigue curve for each temperature based on the six input parameters. After entering the stress amplitude, ANSYS returns the number of cycles to failure (N_f) . The number of cycles to failure is half the number of load reversals $(2N_f)$.

4.5.4 Check 5: the number of cycles to failure (above 538 °C)

ANSYS calculates only the number of cycles to failure based on the six strain-life parameters. The Toolbox gives more information, because it also incorporates the restrictions inserted by the ASME design code: dividing the number of cycles to failure by factor 20 and determining the number of cycles to failure for twice the stress amplitude (via equation (4.5A)). Therefore, the Toolbox uses the fatigue design curve (i.e. the curve including the conservative factors), while ANSYS uses the fatigue curve (i.e. the curve without conservative factors). For the validation in chapter 5, the fatigue curves are compared.

5. Validation of the Toolbox

Calculations have been performed with the Toolbox using certain values for the input parameters temperature, stress and time. The same calculations have been done using the material model in FE software program ANSYS. The results have been compared and based on these comparisons, the validation of the Toolbox is obtained.

5.1 True stress – true strain curves

The maximum number of isothermal true stress – true strain curves that can be constructed in ANSYS is 20. For temperatures in between these specific temperatures, ANSYS calculates the values for the true stress and true strain by linear interpolation. Due to the lack of number of discretisation steps, especially for the true plastic strain, a deviation may occur between the values calculated by ANSYS based on linear interpolation and the values calculated in the Toolbox, that determines these values in a continuous manner. The magnitude of this deviation has been determined for several interpolated values and the values calculated via the MPC model in the Toolbox are within 0.2%. For engineering purposes, this deviation is acceptably small and therefore, both methods (i.e. ANSYS with linear interpolation and Toolbox on a continuous manner) are suitable for determining true stress – true strain relations. Some of the plots are shown in the following figures as an example. The graphs for 20 °C and 600 °C are direct comparisons between the ANSYS values and the Toolbox values. The graph for 775 °C is a comparison between the interpolated results as determined by ANSYS and the direct calculated results from the Toolbox.



Figure 5.1: True stress – true strain curves at 20 °C as calculated by ANSYS (blue diamonds) and as plotted by the MPC model (red line) constructed in the Toolbox.



Figure 5.2: True stress – true strain curves at 600 °C as calculated by ANSYS (blue diamonds) and as plotted by the MPC model (red line) constructed in the Toolbox.



Figure 5.3: True stress – true strain curves at 775 °C as interpolated by ANSYS (blue diamonds) and as plotted by the MPC model (red line) constructed in the Toolbox.

5.2 Maximum allowable stress based on the yield strength

The maximum allowable stress based on the yield strength, which is dominant for temperatures lower than 640 °C, is evaluated in the Toolbox and in ANSYS. Since the equation inserted in both programs is identical, the results for the stress value are as well.

5.3 Time to rupture

Creep damage accumulates at a certain temperature and stress until failure occurs after a certain period, i.e. the time to rupture. The time to rupture calculations have been performed in ANSYS and in the Toolbox and the results have been compared and shown in Table 5A. This is done by calculating the creep rupture strength in the Toolbox in the temperature range of 600 °C – 800 °C for temperature increment steps of 25 °C (i.e. nine different temperatures) and for three different exposure times, namely 2, 10 and 30 years. This resulted in 27 creep rupture strength values for evaluation in ANSYS. Using the default load step increment configuration in ANSYS (i.e. the specimen is divided in 4 segments, twice 0.2 of the total length and twice 0.3 of the total length), the expected time to rupture values are determined using the creep rupture strength values as calculated in the Toolbox and the temperatures at which these strength values have been determined. From the 27 comparisons between chosen time to rupture in the Toolbox and the calculated time to rupture by ANSYS, 26 comparisons resulted in a deviation of 0.25% or less and one was 1.1% (i.e. for 10 years at 775 °C). For engineering purposes and considering the complexity of the equations, it is preferred that the deviation in time to rupture is maximum 0.25%. The cause that one value deviates more than 0.25% is due to the iterative process of non-linear ANSYS calculations. This issue has been solved by carefully choosing a smaller load increment sub-step to obtain a smaller error. By selecting 20 sub-steps of 0.05 each, all deviations are found to be equal to or lower than 0.25%, which proves that the Toolbox and ANSYS are consistent regarding the creep rupture mechanism. The result of the comparisons and deviations are shown in table 5A.

Table 5A: Calculated time to rupture by ANSYS based on the stress values as determined in the Toolbox after time to rupture of 2, 10 and 30 years, and the deviation between the rupture times as set in the Toolbox and as calculated by ANSYS. Except for one value (i.e. for 10 years at 775 °C), these values are determined with the rougher load increment sub-steps.

	Stress to rupture [MPa] as			Time to	Time to rupture [years] as			Deviation of calculated life time by		
	determined in the Toolbox after			calculate	calculated by ANSYS based on			ANSYS compared to the desired life		
		life time		stress	stress to rupture values			time Toolbox		
Temperature	Desire	Desired life time [years]			Desired life time [years]			Deviation [%]		
[°C]	2	10	30	2	10	30	2	10	30	
600	257.09	205.08	175.76	2.001	10.009	30.013	0.04	0.09	0.04	
625	205.39	162.78	138.90	2.002	10.009	30.017	0.08	0.09	0.06	
650	164.09	129.21	109.76	2.002	10.011	30.040	0.10	0.11	0.13	
675	131.10	102.56	86.74	2.002	10.015	30.046	0.10	0.15	0.15	
700	104.73	81.41	68.55	2.003	10.017	30.047	0.17	0.17	0.16	
725	83.67	64.62	54.17	2.004	10.019	30.063	0.20	0.19	0.21	
750	66.85	51.29	42.81	2.004	10.025	30.064	0.18	0.25	0.21	
775	53.41	40.72	33.83	2.004	10.015	30.076	0.18	0.15	0.25	
800	42.67	32.32	26.74	2.004	10.021	30.047	0.20	0.21	0.16	

5.4 Creep strain

Since ANSYS does not differentiate between primary and secondary stresses, certain boundary conditions have been created in ANSYS to investigate the creep strain rate development and adherent the creep strain itself. First, the creep strain rate for the primary stress is determined, which remains constant over time. Therefore, the accumulated creep strain is a multiplication of the constant creep strain rate by the exposure time. This is done in the Toolbox and in ANSYS for seven temperatures (i.e. 650 - 800 °C, with intervals of 25 °C) and for each temperature for three different stresses (i.e. $\frac{1}{3}x$, $\frac{2}{3}x$ and 1x the maximum allowable stress at that temperature according to the ASME code) with an exposure time of 100,000 hours. The results are shown in Table 5B.

As can be seen in Table 5B, the deviation between the Toolbox and ANSYS is maximum 2.0%. Considering the number of temperature-dependent equations out of which the creep strain rate equation exists, each containing its own uncertainty. Also, the creep strain is not the limiting creep criterion up to 800 °C, as the maximum allowable stress is based on creep rupture strength and hence the 1.0% creep strain in 100,000 hours (i.e. 0.01% per 1,000 hours) shall never be reached. The strain rate at temperatures up to at least 750 °C is insignificant, so a deviation of 2.0% is irrelevant. For these reasons, the deviations found are acceptable for the creep strain determination based on a constant primary stress.

			Toolbox	ANSYS	Deviation
Temp.	part x _{Gallow.}	σ	Total creep strain	Total creep strain	
[°C]		[MPa]	[-]	[-]	[-]
650	1/3x 123.90	41.30	9.179E-21	9.317E-21	1.5
650	2/3x 123.90	82.60	7.447E-17	7.580E-17	1.8
650	1x 123.90	123.90	1.441E-14	1.469E-14	1.9
675	1/3x 100.98	33.66	3.449E-17	3.502E-17	1.5
675	2/3x 100.98	67.32	1.031E-13	1.050E-13	1.9
675	1x 100.98	100.98	1.112E-11	1.135E-11	2.0
700	1/3x 81.07	27.02	4.774E-14	4.827E-14	1.1
700	2/3x 81.07	54.05	5.921E-11	6.013E-11	1.6
700	1x 81.07	81.07	3.819E-09	3.882E-09	1.6
725	1/3x 64.15	21.38	2.346E-11	2.355E-11	0.4
725	2/3x 64.15	42.77	1.359E-08	1.369E-08	0.7
725	1x 64.15	64.15	5.618E-07	5.656E-07	0.7
750	1/3x 50.23	16.74	3.969E-09	3.941E-09	-0.7
750	2/3x 50.23	33.49	1.210E-06	1.203E-06	-0.6
750	1x 50.23	50.23	3.434E-05	3.408E-05	-0.7
775	1/3x 39.31	13.10	2.285E-07	2.301E-07	0.7
775	2/3x 39.31	26.21	4.126E-05	4.176E-05	1.2
775	1x 39.31	39.31	8.623E-04	8.731E-04	1.3
800	1/3x 31.35	10.45	4.565E-06	4.541E-06	-0.5
800	2/3x 31.35	20.90	5.500E-04	5.461E-04	-0.7
800	1x 31.35	31.35	9.071E-03	8.996E-03	-0.8

Table 5B: Comparison of the accumulated creep strain after 100,000 hours at several temperatures and several constant (i.e. primary) stresses as calculated by the Toolbox and by ANSYS.

Secondly, the creep strain development for secondary stresses is determined by applying an initial strain and creep causing the stress to diminish (i.e. relax) over time, which in turn decreases the creep strain rate. The accumulated creep strain after a certain amount of time can be determined by an integration of the strain rate equation (i.e. analytical method), see equation (5.4A), or by calculation of the stress and adherent creep strain in small time-steps (i.e. numerical method), see equation (5.4B).

$$\varepsilon_{c,s} = \varepsilon_0 - \left((1-n) \left(\frac{\varepsilon_0^{(1-n)}}{1-n} - B_c \cdot \left(\frac{E}{\mu} \right)^n \cdot \exp\left(\frac{-Q_c}{R \cdot T_K} \right) \cdot t \right) \right)^{\frac{1}{1-n}},$$
(5.4A) and

$$\varepsilon_{c,s} = B_c \cdot \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \cdot \sum_{0}^{t_s = t} \left(\frac{\sigma(t)}{\mu}\right)^n \Delta t \big|_1^{t_s} \qquad , \tag{5.4B}$$

in which $\varepsilon_{c,s}$ the accumulated creep strain after a certain time (t), ε_0 is the initial strain, n is the material exponent for creep, B_c is a temperature-dependent constant, E is the elastic modulus, μ is the shear modulus, Q_c is the activation energy for creep, R is the gas constant, T_K is the temperature in Kelvin, $\sigma(t)$ is the stress at the end of the previous time-step, Δt is the size of the time-step, and t_s is the time at the end of the time-step and t is the time elapsed.

ANSYS and the ASME design code use numeric methods to calculate the diminishing rate of creep strain accumulation. For that reason, the Toolbox has been constructed using the numeric method as well, although the analytic method may be more accurate, since it determines the development of the creep strain accumulation in a continuous manner. For that reason, the accumulated creep strain results from the analytic method, as constructed and calculated in Microsoft Excel, have been evaluated and compared with the results coming from the numeric method performed in the Toolbox (with 480 time-steps, each time-step 1.05x larger than the previous time-step) and those originating from ANSYS (with 1,000 equally sized time-steps). This is done for three different temperatures using two different stresses for each temperature for a period of 100,000 hours (i.e. 3.60⁻10⁸ seconds). Also, the deviations between the results from the Toolbox and those from ANSYS have been compared. The results are shown in Table 5C.

Table 5C: Comparison of the accumulated creep strain due to relaxation (i.e. constant initial strain, decreasing stress) after 100,000 hours at several temperatures determined in an analytic manner, a numeric manner in the Toolbox and a numeric manner in ANSYS.

				Toolbox (480 steps)	ANSYS (1,	Deviation	
Temp.	Time	Stress	ε _{c,s}	ε _{c,s}	Deviation	ε _{c,s}	Deviation	Toolbox
			analytic	numeric	vs. analyt.	numeric	vs. analyt.	vs. ANSYS
[°C]	[s]	[MPa]	[-]	[-]	[%]	[-]	[%]	[%]
700	3.60E+08	120	2.1474E-07	2.1475E-07	0.00	2.1841E-07	1.71	-1.68
700	3.60E+08	240	1.4757E-04	1.4858E-04	0.68	1.4922E-04	1.12	-0.43
750	3.60E+08	75	1.4666E-04	1.4759E-04	0.64	1.4632E-04	-0.23	0.87
750	3.60E+08	150	6.0833E-04	6.0956E-04	0.20	6.0779E-04	-0.09	0.29
800	3.60E+08	40	1.7698E-04	1.7736E-04	0.21	1.7681E-04	-0.10	0.31
800	3.60E+08	80	4.3130E-04	4.3168E-04	0.09	4.3118E-04	-0.03	0.12

The results in Table 5C show that the deviation between the accumulated creep strains as determined by the Toolbox is maximum 0.7% in comparison to the strains as found using the more accurate analytic method. This is an acceptable margin considering the number of variables (i.e. temperature-dependent equations) used to describe the creep strain rate equation.

The accumulated creep strain values as calculated by ANSYS are similar to those determined using the analytic method, although deviations become larger when the temperature decreases or when the stress diminishes. The main reason for the larger deviations using ANSYS is because ANSYS is using large equally sized time-steps. This results in a large error in the beginning of the relaxation process (i.e. where relaxation is the largest and needs to be determined accurately), because the first time-step is already $3.60 \cdot 10^5$ seconds (namely $3.60 \cdot 10^8$ seconds divided by 1,000 time-steps). In comparison, the first time-step in the Toolbox is $2.43 \cdot 10^{-2}$ seconds. Still, values found by ANSYS deviate less than 2% compared to the values calculated using the analytic method, which makes the ANSYS results sufficiently accurate.

Comparison of the creep strain values from the Toolbox and those from ANSYS show maximum 1.7% deviation for the lowest used stress and lowest used temperature, although deviations of 0.3 - 0.4% are more common. For that reason, it may be concluded that the Toolbox and ANSYS results reflect each other well.

5.5 Fatigue for temperatures lower than 425 °C

The fatigue design curve as determined in the Toolbox has been compared with the design curve as constructed by ANSYS. This is done by selecting seven stress values for which the number of cycles to failure has been calculated by both the Toolbox and ANSYS. The deviations between the calculated number of cycles to failure values are less than 0.05%, as can be seen in Table 5D. This is sufficiently accurate for engineering purposes.

Stress	No. of cycle	Deviation	
[MPa]	[·	-]	[%]
	Toolbox		
165	2.074E+08	2.074E+08	0.001
245	1.485E+05	1.485E+05	0.000
325	3.815E+04	3.817E+04	0.049
405	1.711E+04	1.711E+04	0.014
485	9.668E+03	9.669E+03	0.008
565	6.204E+03	6.205E+03	0.007
645	4.317E+03	4.316E+03	-0.007

Table 5D: The number of	f cycles to failure as cal	culated by ANSYS compared	to those	from the Tool	lbox.
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5.6 Fatigue for temperatures higher than 538 °C

300

600

3.98E+05

4.00E+05

1.8357E-03

ANSYS constructs fatigue design curves based on the six strain – life parameters and gives the number of cycles to failure for a given stress amplitude via a black-box calculation algorithm. The Toolbox determines the number of cycles to failure through equations based on the total strain amplitude, which in turn is calculated based on the stress amplitude. The number of cycles to failure originating from the Toolbox and those given by ANSYS are compared for six different temperatures and for seven different stress amplitudes at each temperature. The results are shown in Table 5E.

1[0]-	000				1[0]-	105			
E(T) [MPa] =	1.7334E+05				E(T) [MPa] =	1.6510E+05			
Δσ	εa	N _f (Toolbox)	N _f (ANSYS)	Deviation	Δσ	٤a	N _f (Toolbox)	N _f (ANSYS)	Deviation
[MPa]	[-]	[-]	[-]	[%]	[MPa]	[-]	[-]	[-]	[%]
20	1.1538E-04	1.86E+16	1.00E+15	-	20	1.2114E-04	1.96E+19	1.00E+15	-
50	2.8845E-04	4.49E+12	4.49E+12	0.01	50	3.0284E-04	7.43E+14	7.43E+14	0.00
100	5.7691E-04	8.24E+09	8.24E+09	0.02	100	6.0568E-04	3.36E+11	3.36E+11	0.00
150	8.6555E-04	2.06E+08	2.07E+08	0.21	150	9.0853E-04	3.71E+09	3.71E+09	0.03
200	1.1552E-03	1.52E+07	1.52E+07	0.10	200	1.2116E-03	1.52E+08	1.52E+08	0.23
250	1.4488E-03	2.03E+06	2.03E+06	-0.23	250	1.5158E-03	1.27E+07	1.28E+07	1.21
300	1.7423E-03	3.99E+05	4.01E+05	0.40	300	1.8245E-03	1.72E+06	1.72E+06	0.07
T [°C] =	650				T [°C] =	750			
E(T) [MPa] =	1.6947E+05				E(T) [MPa] =	1.6144E+05			
Δσ	εa	N _f (Toolbox)	N _f (ANSYS)	Deviation	Δσ	εa	N _f (Toolbox)	N _f (ANSYS)	Deviation
[MPa]	[-]	[-]	[-]	[%]	[MPa]	[-]	[-]	[-]	[%]
20	1.1801E-04	1.86E+16	1.00E+15	-	20	1.2388E-04	1.96E+19	1.00E+15	-
50	2.9503E-04	4.49E+12	4.49E+12	0.01	50	3.0971E-04	7.43E+14	7.43E+14	0.00
100	5.9008E-04	8.24E+09	8.24E+09	0.02	100	6.1942E-04	3.36E+11	3.36E+11	0.00
150	8.8529E-04	2.06E+08	2.07E+08	0.19	150	9.2915E-04	3.71E+09	3.71E+09	0.03
200	1.1815E-03	1.52E+07	1.52E+07	0.13	200	1.2391E-03	1.52E+08	1.52E+08	0.22
250	1.4817E-03	2.03E+06	2.03E+06	-0.24	250	1.5501E-03	1.27E+07	1.28E+07	1.18
300	1.7933E-03	3.99E+05	4.00E+05	0.39	300	1.8657E-03	1.72E+06	1.72E+06	0.04
T [°C] =	700				T [°C] =	800			
E(T) [MPa] =	1.6551E+05				E(T) [MPa] =	1.5728E+05			
Δσ	εa	N _f (Toolbox)	N _f (ANSYS)	Deviation	Δσ	εa	N _f (Toolbox)	N _f (ANSYS)	Deviation
[MPa]	[-]	[-]	[-]	[%]	[MPa]	[-]	[-]	[-]	[%]
20	1.2084E-04	1.86E+16	1.00E+15	-	20	1.2716E-04	1.96E+19	1.00E+15	-
50	3.0210E-04	4.49E+12	4.49E+12	0.01	50	3.1791E-04	7.43E+14	7.43E+14	0.00
100	6.0422E-04	8.24E+09	8.24E+09	0.02	100	6.3582E-04	3.36E+11	3.36E+11	0.00
150	9.0650E-04	2.06E+08	2.07E+08	0.19	150	9.5376E-04	3.71E+09	3.71E+09	0.03
200	1.2098E-03	1.52E+07	1.52E+07	0.15	200	1.2719E-03	1.52E+08	1.52E+08	0.22
250	1.5171E-03	2.03E+06	2.03E+06	-0.22	250	1.5912E-03	1.27E+07	1.28E+07	1.15

Table 5E: Comparison of the number of cycles calculated by the Toolbox and those given by ANSYS at several temperatures and stresses.

705

ANSYS maximises the number of cycles to failure to 1.10¹⁵. For that reason, the deviation in number of cycles to failure for a stress amplitude ($\Delta\sigma$) of 20 MPa could not be determined, since the value for the number of cycles to failure as calculated by the Toolbox exceeds 1.10¹⁵ by a (few) order(s) of magnitude.

300

0.39

1.9149E-03

1.72E+06

1.72E+06

0.00

On the left-hand side of Table 5E the numbers of cycles to failure (N_f) are shown for three temperatures, which have been calculated by the Toolbox and by ANSYS with the strain - life parameters for the temperature range 538 – 704 °C. The deviation between the results from the Toolbox and those from ANSYS for these stress amplitudes is 0.4% or less. This deviation value is considered acceptable for engineering purposes, especially considering the large number of cycles involved and the complexity of the fatigue equation.

The numbers of cycles to failure have been calculated by the Toolbox and by ANSYS with the strain – life parameters for the temperature range 704 – 800 °C as well for three different temperatures and the results are shown on the right-hand side of Table 5E. The maximum deviation observed between the Toolbox values and those calculated by ANSYS is about 1.2%, which is acceptable for engineering purposes.

The prediction of the number of cycles to failure can be performed sufficiently accurately for each combination of stress/strain amplitude and temperature between 538 – 800 °C (i.e. the temperature range of interest) using either the Toolbox or ANSYS.

6. Discussions

Some well-considered choices and assumptions have been made in the processing of Alloy 617 data. This is done to unify the data and obtain straight-forward equations, which made it possible to create a Toolbox that functions with only a few input parameters.

6.1 Creep data from creep strain rate

Creep accumulation is partly determined by the creep strain rate, which in turn depends on a number of parameters, like temperature, shear modulus and stress. The creep equations in chapter 2 have been developed for the most ideal material structure, which is a grain size comparable to an ASTM E112, size no. 5 or 6. This grain size is the optimal grain size, since a better creep resistance is often obtained with a larger grain, while a better resistance against fatigue is commonly achieved by more grain boundaries (i.e. a smaller grain). The influence of the grain size on the creep rate is illustrated in figure 6.1 for pure nickel.



Figure 6.1: Deformation mechanism maps for pure nickel with grain size of 1mm (left) and 10 μm (right) [15].

Considering the equation for the creep strain rate, equation 2.3D repeated as equation 6.1A, a difference in creep strain rate ($\dot{\epsilon}_c$) due to a difference in grain size is obtained by adjusting material parameter B_c .

$$\dot{\varepsilon}_c = B_c \left(\frac{\sigma}{\mu}\right)^n \exp\left(\frac{-Q_c}{R \cdot T_K}\right) \tag{6.1A}$$

However, it has been observed that due to careful processing of the material, which involves cold/hotworking and a specific heat treatment, the desired grain size is relatively easy obtainable and reproducible [2-6,64,91]. Also, thermal processes during service may induce grain growth while grain refinement has never been observed. This benefits the creep performance. Therefore, the grain size effect on the creep strain rate is not considered any further.

Creep depends especially on material characteristics, such as grain size, number and distribution of precipitates, carbides and solute atoms, and on temperature and stress. Influences of environmental conditions, such as corrosion and oxidation, are insignificant and they are ignored for that reason.

6.2 Fatigue data

The fatigue data available for Alloy 617 are abundant and the constructed fatigue curves display the fatigue behaviour of Alloy 617 sufficiently accurately. The statistic deviations below the fatigue curves plotted through the data are well within the conservative factors as set by the design code, i.e. the fatigue design curve, which is constructed based on the lowest number of cycles to failure from either by dividing the number of cycles to failure by factor 20 or by taking the number of cycles to failure at the stress amplitude multiplied by factor 2.0 (see paragraphs 2.5.1 and 2.8.2). This is illustrated in figure 6.2, where the data points of Yukawa [60] and INL [4] are combined and fatigue design curves are plotted. The blue diamonds are data points in the temperature range 538 - 704 °C, the red ones are the data points in the temperature range 704 - 871 °C, and the green diamonds reflect those between 871 - 982 °C. The blue solid line is the fatigue design curve at 704 °C, i.e. worst case scenario for the temperature range 538 - 704 °C. The red and green solid lines are the fatigue design curves at $871 \circ C$ and $982 \circ C$ respectively.



Figure 6.2: Empirical data points and fatigue design curves at several temperatures [4, adjusted].

The fatigue design curves could only be constructed for temperatures lower than/equal to 425 °C and higher than/equal to 538 °C. The temperature of 425 °C is set by the ASME design code for their standard fatigue design curve for Alloy 617 and it is most likely chosen as maximum since above this temperature the yield strength becomes significantly temperature-dependent. The fatigue data for 538 °C and higher (up to 982 °C) is available because creep mechanisms become more active in this temperature range. Alloy 617 is considered to be a good material candidate under exposure of combined creep and fatigue conditions and for that reason the fatigue tests have been performed at these elevated temperatures. At temperatures lower than 538 °C, creep influences in Alloy 617 are negligible and other materials are chosen from a cost perspective (i.e. Alloy 617 is too expensive to be used at these lower temperatures where the beneficial characteristics of Alloy 617 are not required). Since Alloy 617 is not considered in the temperature range 425 – 538 °C, fatigue testing has not been performed, hence the lack of data in this temperature range. For the Toolbox, fatigue data in this temperature range may be required because a HRSG system may be 'cycling' on standby in that temperature range to accommodate a 'hot start' (i.e. the system has to go into service soon and quick

and is awaiting the start to go into actual service) or because a tube from a solar receiver may not reach the desired temperature due to bad weather conditions. Also, in case the fatigue is caused by temperature cycling, the ASME design code requires that the fatigue curve at the average temperature is chosen for the determination of the number of cycles to failure, and this temperature can be between 425 and 538 °C. It may be considered to treat the fatigue behaviour in this temperature range by taking the worst-case scenario, which is using the fatigue design curve at 538 °C.

Since the maximum allowable stress is determined for the primary and thermal stress, the chance for low-cycle fatigue failure or ratchetting is limited and especially high-cycle fatigue conditions are considered caused by elastic cycling or elastic shakedown. Nevertheless, the boiler installation has to be able to withstand an extreme situation at least once a year (the so-called level C loadings, maximised at 25 times in total, as described in Appendix F, paragraph F2). This type of loadings may cause strains that might be classified as strains causing low-cycle fatigue. Therefore, the low-cycle fatigue is incorporated in the fatigue assessment and part of both material model in ANSYS and Toolbox.

6.3 Creep-fatigue data

The ASME design code requires that the creep damage and the fatigue damage are determined independently. The interaction between these damage mechanisms is acknowledged by the ASME code though not quantified, because the effect of this interaction cannot be predicted with confidence yet [64]. Therefore, a rather conservative diagram as shown in figure 2.35 is given by the ASME design code. It is possible for a designer to deviate from this diagram (i.e. being less conservative), but this deviation must be supported with valid arguments and calculations to obtain design approval. The uncoupled visco-plastic model as suggested in paragraph 2.7.3.1 would be ideal for the replacement of the ASME diagram. Unfortunately, this diagram cannot be used directly, since it was determined on a single crystal nickel alloy. The absence of grain boundaries has a significant influence on the creep and fatigue characteristics and their interaction. As described before, grain boundaries increase creep due to more diffusion, and their incoherency raises the internal stress in the material which acts as blockers for fatigue crack propagation. Still, a similar uncoupled visco-plastic model is desired for Alloy 617 with grains with ASTM E112 grain sizes of no. 5 or 6 ($45.0 - 63.5 \mum$).

Continuous fatigue tests are used to examine the damage accumulated by fatigue only and fatigue tests with hold-times at the maximum strain are performed to investigate the additional creep effects (i.e. the creep-fatigue interaction). At 950 °C, the cyclic life from the fatigue tests with hold-time seems to be diminished by a factor 3 in comparison to that determined with the continuous cycling tests, regardless the total strain magnitude or hold-time of the fixed strain [6,64,66,78,92]. Also, from the continuous fatigue tests it is concluded that the plastic strain range due to fatigue alone is circa 0.2% lower than the total strain range applied (e.g. the application of a total strain range of 0.6% results in a plastic strain range of 0.4%). The fatigue tests with hold-time showed that the accumulated plastic/creep strain range is approximately 0.1% lower than the total strain range applied (for example, the plastic/creep strain range due to fatigue and combined plastic/creep strain range due to creep-fatigue have been made at three different total strain ranges, namely 0.3%, 0.6% and 1.0%. It may be

concluded that the introduction of a hold-time causes a creep strain range of circa 0.1%, which in turn reduces the cyclic life-time with about a factor 3. This is regardless the duration of the hold-time as can be observed in figure 6.3, left figure.



Figure 6.3: Left figure: Influence of tensile hold-time on number of cycles to failure at 950 °C [92]; Right figure: Rapid stress relaxation during the tensile hold, CF test at 0.3% total strain range [64].

Microscopic examination showed that crack initiation at the surface occurs faster due to oxidation of the grain boundaries at the surface [92,93]. At the same time, cracks develop at the interior grain boundaries, which results in a faster crack propagation due to the linking of these interior grain boundary cracks [92,93]. Also, sub-grains form during cycling due to the tangling of dislocations into walls followed by dislocation climb, reorganisation of dislocations and the formation of well-ordered hexagonal dislocation networks that form the sub-grain-boundaries [93]. The creep-fatigue behaviour is influenced by this combined microstructural evolution and crack propagation and variations in these deformation modes for several strain ranges lead to different macroscopic observations. So is the creep-fatigue deformation for the 0.3% total strain range of Alloy 617 at 950 °C characterised by many interior boundary cracks and a slow propagation of grain boundary surface cracks [94]. This indicates that the creep damage developed during the hold-time has a significant and most probably a dominant influence. The sub-grain formation and the interior boundary cracks lead to cyclic softening on macroscopic scale [92]. The combination of these microscopic processes and the rapid stress relaxation during the hold-time, as shown in the right figure of figure 6.3, makes it feasible that creep and plasticity dominate the time-dependent deformation response in Alloy 617 at 950 °C. Because of the rapid stress relaxation, most damage develops during the first part of the hold-time at maximum tensile strain. The relaxed stress is almost already obtained at the lowest hold-time evaluated, i.e. 180 seconds, as shown in the right figure of figure 6.3. Any additional damage accumulated during the remainder of the hold-time, i.e. up to 600 seconds and 1800 seconds, is negligible in comparison to the damage developed during the first 180 seconds resulting in no or hardly any additional cyclic life reduction for hold-times longer than 180 seconds. This explains that the reduction of cyclic life is a fixed factor for each strain range regardless the hold-time duration.

The same tests have been performed at 850 °C [64]. For the lower total strain range (i.e. 0.3%), the hold-time has a significant influence on the number of cycles to failure as can be observed in the left figure of figure 6.4. Using a hold-time of 3 minutes, the cyclic life is already reduced with a factor of about 5. A hold-time of 30 minutes diminishes the number of cycles to failure even further with approximately factor 10. This effect is less pronounced at higher total strain ranges (i.e. 1.0%), though the focus of this project is especially on the high-cycle fatigue (HCF) parts and therefore on the lower total strain ranges.



Figure 6.4: Left figure: Influence of tensile hold-time on number of cycles to failure at 850 °C [92]; Right figure: Stress relaxation during the tensile hold, CF test at 0.3% total strain range at 850 °C [92].

At 850 °C, the cyclic life decreases with an increase in hold-time because the stress does not fully relax during the strain-controlled hold-time as can be seen in the right figure of figure 6.4. Less stress relaxation for short hold-times means that less creep strain develops during each cycle, and therefore, a larger number of cycles to failure is obtained. This is supported by the microscopic observations, where for continuous cycling (i.e. no hold-time) at a total strain range of 0.3%, transgranular cracks are found that have initiated at the surface, going to a mixed crack configuration for a strain hold-time of 3 minutes and to an intergranular cracking mode for hold-times of 10 minutes and more [92]. The interior intergranular cracks are commonly the result of creep damage. Also, in the specimen with longer hold-times, voids are observed along the interior grain boundaries [92], which are a clear characteristic for creep and which are often crack initiation sites. Another microscopic observation is that similar sub-grains exist just as observed in the 950 °C creep-fatigue specimen [92], though the mobile dislocation density in the sub-grains at 850 °C is significant larger. This higher dislocation density leads to more internal stress, which in turn causes more local plastic deformation and an enlarged creep-void formation, which translate themselves macroscopically in a reduced cyclic life.

The factor of 10 as found for the lower total strain range and the longer hold-time justifies the conservatism of the ASME code diagram for creep-fatigue interaction, since the diagram virtually divides the number of fatigue cycles allowed by a factor 10 when creep mechanisms become active.

Since abundant data concerning creep-fatigue interaction (i.e. the synergic effect of these damage mechanisms combined) is not available, a proposal for a test method to investigate this interaction is described in appendix G to this report.

6.4 Strain-rate sensitivity

Commonly, unique flow stress (i.e. the stress required to continue plastic straining of the material) values are used in the design that are independent of the strain rate. These flow stress values are often determined from conventional tensile tests which are typically performed with a strain rate of 10^{-3} s⁻¹ [95]. The strain-rate sensitivity (m_{srs}) may be determined by strain-rate jump tests or from stress relaxation data. The strain-rate jump test is carried out on a single specimen that is given a fixed strain rate until a steady state flow stress is obtained. Then the strain rate is rapidly increased to the next fixed value and maintained until steady state flow stress is obtained again, etc. The strain-rate sensitivity (i.e. applying a fixed strain on a specimen upon which the stress declines (relaxes)). The derivative of the stress relaxation curve can be taken which reflects the rate of stress change as a function of time. The first method is preferred, because determination of the strain-rate sensitivity from stress relaxation data is found to be inappropriate for stresses above the yield strength [95].

Strain-rate jump tests have been performed for Alloy 617 in the temperature range of 700 - 950 °C and the results are shown in the left figure of figure 6.5 using the strain rates as shown along the horizontal axis of the right figure of figure 6.5 [95].



Figure 6.5: Left: Strain-rate jump test results for Alloy 617 with strain rates as shown in figure 6.6 [95]. Right: log σ_{fl} – log $\dot{\varepsilon}$ plots, including equations for several temperatures [95].

As can be seen from the left figure of figure 6.5, the plots for 700 and 750 °C show serrated flow for one fixed strain rate. Serrated flow is caused by dynamic strain aging, which is the subject of the next paragraph in this chapter. The strain-rate sensitivity is related to the work hardenability [6,95], which is observed from the tensile test results in figure 6.6. When work-hardening is present and one fixed strain rate is used, the flow stress keeps on rising with the accumulation of strain. This is due to the strengthening of the material by plastic deformation, i.e. the formation and restricted movement of dislocations. With restricted movement of dislocations is meant that it is difficult for dislocations to overcome obstacles, like precipitates, solute atoms and grain boundaries.



Figure 6.6: Stress–strain curves as a function of temperature determined at a strain rate of 10⁻⁴ s⁻¹ [95].

For temperatures of 800 °C and above, the sensitivity for the strain rate is evident [5,6,95-98] as can be seen in the left figure of figure 6.5. For these temperatures, no obvious work-hardening is observed, which leads to a steady-state (i.e. constant) stress after some time for a certain strain rate. The absence of work-hardening is most probably caused by thermally activated processes that make it easy for dislocations to overcome obstacles. Examples are faster cross-slip and climb of dislocations. Work-hardening at 800 °C is only present at high strain rates (magnitude of 10^{-2} s⁻¹) [6], which was not tested in the strain-rate jump tests. This is likely because the thermally activated processes are active, but the dislocations do not have sufficient time to overcome the obstacles. When the flow stress (σ_{fl}) depends on the strain rate, their relation may be expressed for a fixed temperature as [6,95,96]:

$$\sigma_{fl} = C_{srs}(\dot{\varepsilon})^{m_{srs}} \qquad [MPa] \qquad , \qquad (6.4A)$$

where C_{srs} is a material constant, $\dot{\epsilon}$ is the strain rate in s⁻¹ and m_{srs} is a dimensionless materialdependent exponent known as the strain-rate sensitivity. Log σ_{fl} – log $\dot{\epsilon}$ plots are constructed from the results of the strain-rate jump tests (see the right figure of figure 6.5). The slope of these plots is the strain-rate sensitivity (m_{srs}) and the material constant C_{srs} is the flow stress in case the strain rate would be 1 (=10⁰). Similar values for the strain-rate dependency have been observed in other literature data [96].

The temperature range considered in this project is up to 800 °C. The applied strain rate can be controlled and strain-rate sensitive behaviour can be avoided in the temperature range up to 800 °C. Also, FE software program ANSYS is not able to incorporate strain rate influences into the stress calculations. Adapting the input parameters for ANSYS would make the calculations complex and time consuming. For these reasons, the strain-rate sensitivity is not considered any further at this moment. In case Alloy 617 is used above 800 °C, the strain-rate sensitivity has to be incorporated into the Toolbox. The Toolbox then requires the strain rate as an additional input parameter to determine the flow stress.

6.5 Dynamic strain aging

Dynamic Strain Aging (DSA), also known as the Portevin-LeChatelier effect, describes the interaction between impurity atoms and the dislocation motion due to applied stress. It demonstrates itself as serrated plastic flow due to locking and releasing of the dislocations by the solute atoms [93,99]. In fatigue testing, DSA is shown as sudden load drops with an adherent shock wave sound. For Alloy 617, DSA lasts during a maximum of 10 cycles for higher total strain ranges (i.e. \geq 1.0%), while for lower total strain ranges, DSA can last for hundreds of cycles before the serrated flow disappears. Also, small strain fatigue favours planar slip of dislocations, while large strain fatigue promotes wavy slip or nonplanar dislocation arrangements [100]. DSA normally prefers planar slip, thus affecting the LCF test results performed at lower total strain ranges to a larger extent [3]. This effect may explain that the data of the lower total strain range LCF tests show more scatter [93,99].

The effects of DSA are not incorporated in the stress development because DSA occurs in the plasticstrain region [64]. The allowable stress is commonly below the yield strength, so DSA is not likely to occur for the design conditions for primary stresses. Though, it is noteworthy that the solute atoms in Alloy 617 can cause DSA in the material when a stress is applied.

6.6 Pressurised tubes versus conventional tests

Conventional test specimens are commonly taken from plate material and tested in a uniaxial direction. Tests with tubes that are pressurised internally have been performed to examine the influence of applying a bi-axial stress (i.e. an internal pressure produces an axial and a hoop stress instead of solely a uniaxial stress) and the effect of the thickness of the wall. Idaho National Laboratory (INL), Boise State University (BSU) and Argonne National Laboratory (ANL) have performed tests to illustrate these influences [19] and the Larson-Miller Parameter data are plotted in figure 6.7.



Figure 6.7: Larson-Miller plot for Alloy 617 uniaxial specimens and pressurised tubes [19].

From figure 6.7 can be observed that the data for the pressurised tubes fall below the trend-line of those of the conventional uniaxial tests. Hence, for a fixed temperature, a bi-axial stress in pressurised tubes causes a time to rupture that is shorter than that for a similar uniaxial stress in conventional test specimens.

A similar observation is done for the Monkman-Grant relation between the minimum creep rate and the time to rupture (see figure 6.8). The pressurised tubes tested by INL were 2 mm thick, while those tested by BSU were 1 mm thick. Since the time till rupture is important for the lifetime estimation in engineering design, the minimum strain rate has to be determined/adjusted for pressurised tubes. Roughly, to achieve a certain lifetime (i.e. time to rupture) the allowable minimum creep rate in pressurised tubes should be about a factor 100 less than the minimum creep rate as determined from conventional uniaxial test specimen.



Figure 6.8: Monkman-Grant plot of time to rupture as a function of minimum creep rate for uniaxial specimens and pressurised tubes [19].

Microscopic examination showed that cracks have developed parallel to the axial direction of the tube, hence perpendicular to the hoop stress, which is the maximum stress as expected. Usually these cracks are surface cracks with a 45° angle relative to the tube radius. Above a radial strain of approximately 5%, the cracks that have nucleated at the surface of the tube starts to open rapidly with an increase in strain [19]. This occurs at a much faster rate than observed in conventional uniaxial test specimens. The pressurised tubes have achieved final failure when they are not able to retain the pressure. This occurs damage in front of one of the surface cracks are linking.

The shorter time to rupture for pressurised tubes is caused by the two-component normal stress (i.e. bi-axial stress) which results in more damage than is observed for the uniaxial tests. These enhanced stresses/loads are already considered by the FE software program. Also, the maximum allowable stress based on the creep rupture data is 80% of the minimum creep rupture strength, as described in paragraph 2.8, partly to account for geometrical and stress distribution influences. Therefore, the difference between uniaxial test results and pressurised tube test results is not considered any further.

6.7 Material long-term exposure

Another challenge, besides surface-oxidation resistance, is the complexity of the aging effects in Alloy 617, because observations and predictions are not consistent about which precipitates form at certain temperatures. The aging effects depend on precipitation kinetics and coarsening processes. Precipitates may form during initial exposure at high temperatures. On the other hand, a large part of the γ' -precipitates dissolves after long-time exposure at service temperature and Alloy 617 depends on solid-solution strengthening only after long exposure time. Aging at 700 – 750 °C results in a tensile and impact property reduction. These effects are less pronounced at higher temperatures [4].

6.8 Other Materials

Alloy 617 has been selected for the next generation HRSG pressure retaining components and for solar boiler tubes, since this material shows the most promising performance at high temperatures in service conditions that may cause creep-fatigue damage. In the design code considered (i.e. the ASME code), the assessment for the application of Alloy 617 in boiler components has advanced quite far already. Alloy 617 has been altered by some material manufacturers and other materials have been developed, which could withstand the same service conditions just as well and perhaps even better than Alloy 617. Unfortunately, insufficient data is available to incorporate these materials into the design for pressurised components with confidence and these materials are not officially recognised by the design code yet. Some of these materials are Alloy 617 mod., Alloy 617B, Alloy 740H and Alloy 230.

7. Conclusions

Alloy 617 is a precipitate-strengthened and solute-strengthened nickel-based material, which is proven to withstand yielding, creep and fatigue adequately at temperatures up to 800 °C during long-term exposure while a good heat transfer over the component's wall is maintained. Other advantages are that abundant data are available for this material and that it is an approved material by the used design code.

The material properties of Alloy 617 are equated into temperature-stress-time-dependent equations. The material property equations for strength are based on the lower-bound plots of available data. This results in the minimum yield strength, minimum tensile strength and minimum creep rupture strength as a function of temperature. The damage mechanisms fatigue, ratchetting, plastic shakedown and fast creep strain rate damage are equated as a function of temperature, applied stress, exposure time and/or number of expected load/thermal cycles based on best-fit plots drawn through the available data. Influences on these material properties and damage mechanisms, such as creep-fatigue interaction, strain-rate sensitivity, dynamic strain-aging, long-term high-temperature exposure and multi-axial loading instead of uniaxial loading, have been evaluated and discussed. It is concluded that the strength and damage mechanism equations can be used unconditionally in the temperature range up to 800 °C.

A material model is built in FE software program ANSYS. Besides the minimum strength and damage accumulation equations, material property equations such as coefficient of thermal expansion, thermal conductivity and specific heat are incorporated. Simulations have been run using boundary conditions on the input parameters temperature, stress and time, and the results are as expected and satisfactory. This allows the direct use of Alloy 617 into the boiler component design. Also, some of the design code limits are built into this material model to ensure a safe and reliable design, but a few of these limits could not be incorporated due to restrictions in the FE software. Only one data gap exists, which is the fatigue data in the temperature range 425 - 538 °C.

A Toolbox is constructed as well, that performs the same calculations as the material model in ANSYS, but additionally identifies the dominant damage mechanism, considers all design code's conservative factors and limits and provides a fast overview of all relevant material and design code criteria. Other advantages of the Toolbox over ANSYS are that the Toolbox determines the true stress – true strain relation in a continuous manner where ANSYS uses linear interpolation to find in-between values and that the Toolbox calculates the accumulated creep strain more accurately. For these reasons, the analysis of the design calculations can be done in the Toolbox rather than in ANSYS. The Toolbox is validated by FE simulations and can easily be extended with other materials next to Alloy 617.

References

- [1] Website from NEM Energy BV, consulted April 8th, 2015. http://www.nem-group.com/EN/products/hrsgs/1/
- [2] T. Yonezawa, Nickel Alloys: Properties and Characteristics, Tohoku University, Japan, Elsevier, 2012.
- [3] X. Chen, *High Temperature Creep-Fatigue Behavior of Alloy 617 and Alloy 230*, Doctor thesis, University of Illinois, Urbana, Illinois (USA), 2012.
- [4] Idaho National Laboratory (INL), *NGNP High Temperature Materials White Paper*, INL/EXT-09-17187 rev.1, Aug. 2012.
- [5] J.K. Wright, L.J. Carroll, C.J. Cabet, Ti. Lillo, J.K. Benz, J.A. Simpson, W.R. Lloyd, J.A. Chapman, R.N. Wright, *Characterization of Elevated Temperature Properties of heat Exchanger and Steam Generator Alloys*, Nuclear Engineering and Design, Vol. 251, p. 252-261, 2011.
- [6] T.C. Totemeier, H. Tian, *Creep-fatigue-environment interactions in INCONEL 617,* Idaho National Laboratory (USA), ed. Elsevier, Materials Science & Engineering A 468 470, p. 81 87, 2007.
- [7] ASME Boiler & Pressure Vessel Code, Section II Part B: Non Ferrous Material Specifications, ASME, 2013.
- [8] W. Ren, T. Totemeier, M. Santella, R. Battiste, D.E. Clark, Status of Testing and Characterization of CMS Alloy 617 and Alloy 230, ORNL/TM-2006-547, Oak Ridge National Laboratory, 2006.
- [9] V.K. Vasudevan, Mechanisms Governing the Creep behavior of high temperature alloys for Generation IV Nuclear Energy Systems, University of Cincinnati (USA), Project No. 09-776, in collaboration with Idaho National Laboratory and Oak Ridge National Laboratory, April 2015.
- [10] W.D. Callister Jr, *Materials Science and Engineering, An Introduction, 4th edition,* Department of Metallurgical Engineering, University of Utah (USA), John Wiley & Sons, Inc., 1997.
- [11] W-G. Kim, M.W. Kim, Y-W Kim, An Improved Methodology for Obtaining the Initial Loading Strain in Creep of Alloy 617, Korea Atomic Energy Research Institute, Daejeon, Korea, in Transactions of the Korean Nuclear Society Spring Meeting, Gwangju, Korea, May 2013.
- [12] N. Lybeck, T.-L. Sham, Development of yield and tensile strength design curves for Alloy 617, International Workshop on Structural Materials for Innovative Nuclear Systems, INL/CON-12-28782 (Preprint), Idaho National Laboratory (USA), October 2013.
- [13] ASME Boiler & Pressure Vessel Code, Section II Part D: Properties (metric) Materials, ASME, 2011.
- [14] T.-L. Sham, D.R. Eno, K.P. Jensen, Treatment of High Temperature Tensile Data for Alloy 617 and Alloy 230, Proceedings of 2008 ASME Pressure Vessels and Piping Division Conference, PVP2008-61128, Chicago (IL), USA, 2008.
- [15] <u>http://engineering.dartmouth.edu/defmech/</u>, consulted January 19th, 2016.
- [16] API 579-1/ASME FFS-1, "*Fitness-For-Service*", American Society of Mechanical Engineers and American Petroleum Institute, June, 2007.
- [17] ASME III, Rules for construction of Nuclear Facility Components, American Society of Mechanical Engineers, 2013.
- [18] ASME VIII, Div. 2, *Rules for construction of Pressure Vessels, Alternative Rules,* ASME Boiler and Pressure Vessel Code, 2013.
- [19] J.K. Wright, R.N. Wright, *Creep Rupture of Pressurized Alloy 617 Tubes*, INL/EXT-13-30251, Idaho National Laboratory, Sept. 2013.
- [20] C. Zener, J.H. Hollomon, *Effect of Strain Rate upon Plastic Flow of Steel*, Journal of Applied Physics 15, p. 22-32, 1954.
- [21] V. Knezevic, A. Schneider, c. Landier, Creep Behaviour of Thick-Wall Alloy 617 Seamless Pipes for 700 °C Power Plant Technology, 6th International Conference on Creep, Fatigue and Creep-Fatigue Interaction [CF-6], SciVerse ScienceDirect, Procedia Engineering 55, Elsevier, p. 240 – 245, 2013.
- [22] J.A. Charles, F.A.A. Crane, J.A.G. Furness, *Selection and use of engineering materials*, Butterworth-Heinemann, Oxford, 2001.
- [23] R. W. Hertzberg, *Deformation and fracture mechanics of engineering materials*, 4th edition, 1995, Published by: Wiley.
- [24] F. C. Monkman and N. J. Grant, *Proc. ASTM 56*, 1956.
- [25] X.Wen, Creep behaviour of High Temperature Alloys for Generation IV Nuclear Energy Systems, University of Cincinnati, USA, 2014.
- [26] O.D. Sherby, P.M. Burke, *Mechanical Behavior of Crystalline Solids at Elevated Temperature*, Progress in Materials Science 13, p. 323-390, 1968.
- [27] A.K. Mukherjee, J.E. Bird, et al., *Experimental Correlations for High-Temperature Creep*, ASM Transactions Quarterly 62, p. 155-179, 1969.
- [28] J. Weertman, P. Shahinian, Creep of Polycrystalline Nickel, Journal of Metals 8, p. 1223-1226, 1980.

- [29] Special Metals, Inconel Alloy 617, 2005 http://www.specialmetals.com/documents/Inconel%20alloy%20617.pdf .
- [30] K. Schneider, W. Hartnagel, et al., *Creep behaviour of materials for high-temperature reactor application*, Nuclear Technology, V.66, p.289, 1984.
- [31] R. Lagneborg, *Bypassing of Dislocations Past Particles by Climb Mechanism*, Scripta Metallurgica 7, p. 605-614, 1973.
- [32] R.S.W. Shewfelt, L.M. Brown, *High-temperature Strength of Dispersion-hardened Single Crystals II Theory,* Philosophical Magazine 35, p. 945-962, 1977.
- [33] H.E. Evans, G. Knowles, *Threshold stress for Creep in Dispersion-Strengthened Alloys,* Metals Science 14, p. 262-266, 1980.
- [34] P.W. Davies, G. Nelmes, et al., *Stress-Change Experiments During High-Temperature Creep of Copper, Iron, and Zinc,* Metal Science 7, p. 87-92, 1973.
- [35] J.D. Parker, B. Wilshire, *The Effect of a Dispersion of Cobalt Particles on High-Temperature Creep of Copper*, Metal Science 9, p. 248-252, 1975.
- [36] E. Artz, D.S. Wilkinson, *Threshold Stresses for Dislocation Climb Over Hard Particles: The Effect of an Attractive Interaction*, Acta Metallurigica 34, p. 1893-1898, 1986.
- [37] Y. Han, M.C. Chaturvedi, A Study of Back Stress During Creep Deformation of a Superalloy Inconel 718, Materials Science and Engineering 85, p. 59-65, 1987.
- [38] M. McLean, On the Threshold Stress for Dislocation Creep in Particle Strengthened Alloys, Acta Metallugica 22, p. 545-556, 1985.
- [39] T. Lillo, High Temperature Alloy 617 Properties for Engineering Design: Creep, Fatigue & Creep-Fatigue, Idaho National Laboratory (USA), Technical Meeting on High-Temperature Qualification of High Temperature Gas Cooled Reactor Materials, June 2014.
- [40] W.J. Evans, G.F. Harrison, Anelastic Deformation and Stress Reduction Experiments During Creep, Scripta Metallurgica, Vol.9, p. 239-246, 1975.
- [41] P.J. Henderson, M. McLean, *Microstructural Contributions to Friction Stress and Recovery Kinetics During Creep of the Nicel-base Superalloy IN738LC*, Acta Metallurgica, V.31, p. 1203-1219, 1983.
- [42] R. Kozar, A. Suzuki, et al., *Strengthening mechanisms in Polycrystalline Multimodal Nickel-base Superalloy,* Metallurgical and Materials Transactions A, V.40A, p. 1588-1603, 2009.
- [43] E. Arzt, E. Göhring, A model for dispersion strengthening of ordered intermatallics at high temperatures, Max-Planck-Institut, University of Stuttgart, Germany; Acta Materiallia, Vol. 46, No. 18, p. 6575-6584, 1998.
- [44] E. Arzt, D.S. Wilkinson, *Threshold stresses for dislocation climb over hard particles: The effect of an attractive interaction*, Max-Planck-Institut, Stuttgart, Germany; Acta Materiallia, Vol. 34, No. 10, p. 1893-1898, 1986.
- [45] G.I. Taylor, *Plastic Strain in Metals*, Journal of the Institute of Metals, Vol. 62, p. 307-324, 1938.
- [46] R.E Stoller, S.J. Zinkle, On the relationship between uniaxial yield strengt hand resolved shear stress in polycrystalline materials, Journal of Nuclear Materials, 283-287, Section 4. Mechanical behavior of irradiated materials, p. 349-352, Elsevier, 2000.
- [47] A.J. Ardell, *Precipitation Hardening*, Metallurgical Transactions A, V.16A, p.2131-2165, 1985.
- [48] J. Rösler, E. Arzt, *The Kinetics of Dislocations Climb over Hard Particles I. Climb without attractive particledislocation interaction,* Acta Metallurgica, V.36, p1043-1051, 1988.
- [49] M.E. Krug, D.C. Dunand, Modelling the Creep Threshold Stress due to Climb of Dislocation in the Stress Field of a Misfitting Precipitate, Acta Metallurgica, V.59, p.5125-5134, 2011.
- [50] H. Nickel, F. Schubert, H.J. Penkalla, H.H. Over, *Mechanical Design Methods for High Temperature Reactor Components,* Nuclear Engineering and Design 76, p. 197-206, Amsterdam (NL), 1983.
- [51] S. Kwofie, H.D. Chandler, *Fatigue life prediction under conditions where cyclic creep-fatigue interaction occurs,* International Journal of Fatigue V. 29, p. 2117-2124, Elsevier, 2007.
- [52] O.H. Basquin, The exponential law of endurance test, Proc. ASTM, p. 625-630, 1910.
- [53] L.F. Coffin Jr., A note on low cycle fatigue laws, Journal of Materials, p. 388-402, 1970.
- [54] S.S. Manson, *Behaviour of materials under conditions of thermal stress*, Report 1170, National Advisory Commission on Aeronautics, 1954.
- [55] N.E. Dowling, *Mechanical behaviour of materials: engineering methods for deformation, fracture and fatigue,* Englewood Cliffs, Prentice-Hall, p. 673, 1993.
- [56] R. Wright, *VHTR Materials Overview*, DOE-NE Materials Crosscut Coordination Meeting, Idaho National Laboratory, July 2013.
- [57] O.K. Chopra, W.J. Shack, *Review of the margins for ASME Code fatigue design curve effects of surface roughness and material variability*, Argonne National Laboratory (USA), 2003.

- [58] W.E. Cooper, *The initial scope and intent of the Section III Fatigue Design Procedure*, Technical information from Workshop on Cyclic Life and Environmental Effects in Nuclear Applications, 1992.
- [59] L. Anand, D.M. Parks, 2.002 Mechanics and Materials II, Spring 2004, Massachusetts Institute of Technology, Department of Mechanical Engineering, Cambridge, Massachusetts (USA), 02139.
- [60] S. Yukawa, *Elevated Temperature Fatigue Design Curves for Ni-Cr-Co-Mo Alloy 617*, The 1st JSME/ASME Joint International Conference on Nuclear Engineering, Tokyo, 1991.
- [61] H. Hübel, A. Willuweit, J. Rudolph, R. Ziegler, H. Lang, K. Rother, S. Deller, *Performance study on the simplified theory of plastic zones and the Twice-Yield method for he fatigue check,* International Journal of Pressure Vessels and Piping, V116, p. 10-19, 2014.
- [62] J. Bree, Elastic-plastic behaviour of thin tubes subjected to internal pressure and intermittent high-heat fluxes with application to fast-nuclear-reactor fuel elements, Journal of Strain Analysis, V2 N3, 1967.
- [63] <u>https://www.researchgate.net/figure/260594392_fig1_Figure-11-Bree-Diagram-of-pressurized-thin-wall-tube-under-thermal-loading-4345</u>
- [64] L. Carroll, M. Carroll, *Creep-Fatigue Behavior of Alloy 617 at 850 and 950 °C*, INL/EXT-13-28886, Rev. 1, Idaho National Laboratory, USA, May 2013.
- [65] Y.L. Lu, L.J. Chen, G.Y. Wang, M.L. Benson, P.K. Liaw, S.A. Thompson, J.W. Blust, P.F. Browning, A.K. Bhattacharya, J.M. Aurrecoechea, and D.L. Klarstrom, *Hold time effects on low cycle fatigue behaviour of Haynes 230 superalloy at high temperatures*, Materials Science and Engineering A, vol. 409, p. 282 291, 2005.
- [66] P. Rodriguez, K.B.S. Rao, *Nucleation and growth of cracks and cavities under creep-fatigue interaction*, Progress in Material Science 37, p. 403 480, 1993, Elsevier.
- [67] W.J. Plumbridge, E.G. Ellison, *Low-cycle-fatigue behaviour of superalloy blade materials at elevated temperature,* Journal of Materials Science & Technology 3, p. 706 – 714, 1987.
- [68] T. Tinga, W.A.M. Brekelmans, M.G.D. Geers, *Time-incremental creep-fatigue damage rule for single crystal Ni-base superalloys,* Materials Science and Engineering A 508 (2009), p. 200 208.
- [69] T. Tinga, W.A.M. Brekelmans, M.G.D. Geers, *Multiscale Modelling of Single Crystal Superalloys for Gas Turbine Blades*, Philosophical Magazine, vol. 88, p. 3793–3825, 2008.
- [70] S.J. Moss, G.A. Webster, E. Fleury, *Creep Deformation and Crack Growth behavior of a Single-Crystal Nickel-Base Superalloy*, Metallurgical and Materials Transactions A, vol. 27A, p. 829 837, 1996.
- [71] A. Epishin, T. Link, U. Bruckner, P.D. Portella, *Kinetics of the Topological inversion of the* 2/2'-microstructure during creep of a nickel-based superalloy, Acta Materialia, vol. 49, p. 4017–4023, 2001.
- [72] K. Serin, G. Gobenli, G. Eggeler, On the influence of stress state, stress level and temperature on gamma-channel widening in the single crystal superalloy CMSX-4, Material Science and Engineering A, vol. 387–389, p. 133–137, 2004.
- [73] C. Carry, J.L. Strudel, Apparent and effective creep parameters in single crystals of a nickel base superalloy II. Secondary creep, Acta Metallurgica, Vol. 26, p. 859–870, 1978.
- [74] N. Zhou, C. Shen, M.J. Mills, Y. Wang, *Mechanisms of γ' rafting in single crystal Ni-base Superalloys A simulation study*, AFOSR under MEANS 2, Ohio State University.
- [75] D.Y. Li, L.Q. Chen, *Computer simulation of morphological evolution and rafting of* γ' *particles in Ni-based Superalloys under applied stresses*, Scripta Materialia, Vol. 37, No. 9, p. 1271-1277, Elsevier Science Ltd, USA 1997.
- [76] B. Fedelich, A microstructural model for the monotonic and the cyclic mechanical behavior of single crystals of superalloys at high temperatures, International Journal of Plasticity, vol. 18, p. 1–49, 2002.
- [77] A. Shyam, W.W. Milligan, Acta Materialia, A model for slip irreversibility, and its effect on the fatigue crack propagation threshold in a nickel-base superalloy, vol. 53, p. 835–844, 2005.
- [78] K. B. S. Rao, H. Meurer, H. Schuster, *Creep-fatigue interaction of Inconel 617 at 950 °C in simulated nuclear reactor helium*, Material Science and Engineering A, vol. 104, p. 37 51, 1988.
- [79] J. R. Davis and ASM International Handbook Committee, Heat-Resistant Materials, Materials Park, Ohio, ASM International, 1997.
- [80] J. Lecompte-Beckers, F. Schubert, *Materials for Advanced Power Engineering 1998,* Forschungszentrum Jülich, 1998.
- [81] A. Palmgren, Zeitung für Vereinigten Deutschen Ingenieurs, vol. 68, p. 339 341, 1924.
- [82] M.A. Miner, Journal of Applied Mechanics, p. A159 A164, 1954.
- [83] S. Majumdar, P.S. Maiya, Journal of Engineering Materials Technology, vol. 102, p. 159 167, 1980.
- [84] M. Satoh, E. Krempl, *Material behaviour at elevated temperatures and components analysis*, p. 71 79, related to ASME.
- [85] J.K. Wright, Strain Rate Sensitivity of Alloy 617, Very High Temperature Reactor (VHTR) R&D 4th Annual Technical Review Meeting, 2011.

- [86] D.K. Fork, J. Fitch, S. Ziaei, R.I. Jetter, *Life Estimation of Pressurized-Air Solar-Termal Receiver tubes*, Journal of Solar Energy Engineering, SOL-12-1058, 2012.
- [87] API 579-1/ASME FFS-1, "Fitness-For-Service", American Society of Mechanical Engineers and American Petroleum Institute, June, 2007.
- [88] VDM Alloy 617, Nicrofer 5520 Co, VDM Metals, Material Data Sheet no. 4019, January 2005 edition.
- [89] ASTM A370-11A, Standard Test Methods and Definitions for Mechanical Testing of Steel Products, ASTM International, 2011.
- [90] ASTM E8, Standard Test Methods for Tension Testing of Metallic Materials, ASTM International, 2011.
- [91] G. R. Halford, *Evolution of creep-fatigue life prediction models*, Winter Annual Meeting of the American Society of Mechanical Engineers, p. 43-57, December 1991.
- [92] L. Carroll, Progress report on long hold-time creep-fatigue of Alloy 617 at 850 °C, INL, October 2015.
- [93] M.C. Carroll, L.J. Carroll, *Developing dislocation subgrain structures and cyclic softening during high-temperature creep-fatigue of a nickel alloy,* Metallurgical and Material Transactions, V.44, No. 8, p.3592-3607.
- [94] L. Carroll, C, Cabet, M. Carroll, R. Wright, *The development of microstructural damage during high temperature creep-fatigue of a nickel alloy*, International Journal of Fatigue, V.47, p.115 125.
- [95] J.K. Wright, J.A. Simpson, R.N. Wright, L.J. Carroll, T.L. Sham, *Strain rate sensitivity of Alloys 800H and 617*, PVP2013, INL/CON-12-27810 Preprint, July 2013.
- [96] R.A. Mulford, U.F. Kocks, *New Observations on the Mechanisms of Dynamic Strain Aging and of Jerky Flow,* Acta Metallurgica, vol. 27, p. 1125, 1979.
- [97] H.P. Meurer, G.K.H. Gnirss, W. Mergler, G. Raule, H. Schuster, G. Ullrich, *Investigations on the fatigue behavior of high-temperature alloys for high-temperature gas-cooled reactor components,* Nuclear Technology, Vol. 66(2), p. 315-323, 1984.
- [98] M.E. Abd El-Azim, K.E. Mohamed, F.H. Hammad, *The deformation characteristic of Alloy 800H and Alloy 617*, Mechanics of Materials, Vol.14, p.33-46, 1992.
- [99] M. S. Rahman, G. Priyadarshan, K. S. Raja, C. Nesbitt and M. Misra, *Characterization of high temperature deformation behavior of Inconel 617*, Mechanics of Materials, vol. 41, p. 261-270, 2009.
- [100] M. Gell and G. R. Leverant, *Mechanisms of high-temperature fatigue, Fatigue at Elevated Temperatures*, ASTM STP 520, p. 37-67, 1973.

Appendix A – Creep

Creep strain increases over time under the application of a static load or stress until failure occurs. The strain – logarithmic time curve that can be drawn from these data can be divided in three regimes (see figure A1). After load initiation (at strain ϵ_0), the true strain rate ($\dot{\epsilon}_s = d\epsilon / dt$) decreases with time to a steady-state value during regime I. During this stage, strain-hardening mechanisms develop which make the material stronger, hence resulting in the diminishing strain rate. Strain hardening is caused by sub-grain formation associated with the rearrangement of dislocations [A1]. In regime II, the true strain rate is in a steady-state condition, in which the hardening and the softening processes are in equilibrium and the sub-microstructure is stable. Softening occurs due to recovery processes like thermally activated cross-slip and edge dislocation climb. The creep rate increases during regime III until failure occurs. This is due to a number of metallurgical instabilities, that result in the loss of balance between the hardening and softening mechanisms. Some of these instabilities are localised necking, micro-void formation, precipitation of brittle second-phase particles, dissolving of second phases that originally contributed to the alloy strengthening, recrystallisation of the strain-hardened grains, intercrystalline fracture and corrosion [A1].



Figure A1: creep curve (typically) showing three regimes [A1].

A1. Relation between rupture time and steady-state strain rate

For long-life material applications, like in a HRSG, the minimum creep rate is the key material response for a given stress and temperature. This information is obtained by performing creep tests into regime II, since the creep rate $\dot{\epsilon}_s$ in this regime is in a steady-state. The accuracy of $\dot{\epsilon}_s$ increases with time in regime II, which is the downside of most creep tests, since these tests are either insufficient accurate or time consumptive. The best creep data available is the data evolved from samples that were taken from an installation that had been in service for a certain period. Over the years, an equation has been proposed, which agrees with and summarises the findings of the combined researches [A1]:

$$\epsilon_t = \epsilon_{0_t} + \beta t^m + \dot{\epsilon}_S t \qquad , \qquad (A1)$$

in which:

ϵ_t	= true strain over time
ϵ_{0_t}	= instantaneous true strain due to the application of a static load
β	= time-independent constant for parabolic creep
t	= time
т	= time-independent material constant ($0 \le m \le 1$)
$\dot{\epsilon}_s$	= true strain rate (linear/steady-state)

The material constant 'm' is independent on time, but does depend on stress and temperature: m increases with an increase in temperature and/or stress. At low temperatures m is zero. \dot{e}_s is linear since this unit is referring to the strain rate in regime II: hardening and softening mechanisms are in steady-state.

Monkman and Grant found a relation between the rupture time (t_R) and the true strain rate [A1, A2], which is often used for determining life expectation of an installation.

$$\log t_R + m \log \dot{\epsilon}_S = B \tag{A2}$$

For several alloys, Monkman and Grant found values for the material constants m and B, but the main conclusion is that the rupture time and the true strain rate have an inversely proportional behaviour.

A2. Steady-state strain rate

Although \dot{e}_s is linear, it is dependent on the temperature (*T*), applied stress (σ), creep strain (ϵ), and material constants, which involve intrinsic lattice properties (m_1), like elastic shear modulus (*G*) and crystal structure, and metallurgical parameters (m_2), such as (sub)grain size, thermo-mechanical history and energy of stacking faults [A1]. The metallurgical parameters (m_2) are by themselves also dependent on temperature, stress and strain.

For a stable microstructure and a fixed applied stress, the strain-rate – temperature relation is expressed with the Zener-Hollomon parameter (Z) [A1]:

$$Z = \dot{\epsilon_S} \exp\left(\frac{\Delta H_C}{RT_K}\right) \qquad , \tag{A3}$$

in which:

 $\begin{array}{ll} \Delta H_C & = \mbox{activation energy of the creep mechanism} \\ R & = \mbox{gas constant} \\ T_K & = \mbox{temperature (in K)} \end{array}$

Since Z is considered constant, the result of equation (A3) is that for a certain creep mechanism (ΔH) the relation between log $\dot{\epsilon}_S$ and 1/T is linear inversely proportional. Hence the strain rate increases with an increase in temperature.
The relation between log $\dot{\epsilon}_S$ and 1/T is non-linear in case multiple creep mechanisms are present which are dependent and concurrent of each other. The slowest creep mechanism controls $\dot{\epsilon}_S$ and the total strain rate $\dot{\epsilon}_{S,tot}$ is [A1]:

$$\frac{1}{\dot{\epsilon}_{S,tot}} = \frac{1}{\dot{\epsilon}_1} + \frac{1}{\dot{\epsilon}_2} + \dots + \frac{1}{\dot{\epsilon}_n} \qquad , \qquad (A4)$$

where 'n' is the number of active creep mechanisms. When the creep mechanisms are independent from one-another, the fastest mechanism dominates the creep process:

$$\dot{\epsilon}_{S,tot.} = \dot{\epsilon}_1 + \dot{\epsilon}_2 + \dots + \dot{\epsilon}_n \tag{A5}$$

Since there are usually more creep mechanisms present in materials, some dependent and others independent of one-another, the activation energy and thus the total creep rate varies with stress and temperature. Therefore, stress and temperature (also) dictate the dominant creep mechanism.

A2.1 Influence of melting temperature

The creep activation energy of metals has the tendency to increase with temperature until the homologous temperature (T_h) of 0.5. The T_h is the ratio between the actual temperature T (in K) and the melting temperature T_{melt} (also in K). For $T_h \ge 0.5$, the creep activation energy seems to reach a constant value, which is in magnitude equal to the activation energy for self-diffusion (ΔH_{SD}). Through the equations for diffusion and considering the crystal structure and the valence of the material, the self-diffusion activation energy can be expressed as [A1]:

$$\Delta H_{SD} = R \cdot T_{melt}(K_0 + V) \qquad , \qquad (A6)$$

in which:

R	= gas constant
K ₀	= parameter dependent on crystal structure/packing density
	(BCC = 14, FCC and HCP = 17, diamond cubic = 21)
V	= valence of metal

Thus the ΔH_{SD} / ΔH_C increases with an increase in melting temperature, valence, packing density, and/or degree of covalence. So, a high melting temperature leads to a higher creep activation energy and thus in a better creep resistance.

A2.2 Stress influence

The stress (σ) and the shear modulus (G) have significant roles in the creep behaviour of a material, since these parameters have a semi-empirical relationship with the true strain rate ($\dot{\epsilon}_s$) [A3] :

$$\log \frac{\dot{\epsilon_S} k_B T}{DGb} = \log A + n \log \left(\frac{\sigma}{G}\right) \qquad , \tag{A7}$$

where:

 k_B = Boltzman's constantD= diffusivityb= Burgers vectorA, n= material constant that are empirically determined

The effect of stress at $T_h > 0.5$ is empirically determined:

- low σ : viscous creep, since n = 1, so $\dot{\epsilon}_s \div \sigma$
- intermediate σ : power-law creep, since n = 4-5, so $\dot{\epsilon}_s \div \sigma^n$
- Very high σ : exponential creep, so $\dot{\epsilon}_s \div e^{\alpha\sigma}$ for $\alpha\sigma > 1.2$ (empirically) and α is a constant.

A3. Creep mechanisms

The mechanisms causing creep can roughly be divided into diffusional and dislocation movement creep. The dominant mechanism is for the greater part determined by the temperature and applied stress. Commonly, creep is considered having a significant role at temperatures $\geq 0.5 T_{melt}$, which is referred to as 'high temperature' in the remainder of this chapter.

A3.1 Low stresses and high temperatures

At low stresses and high temperatures, diffusional creep is often the dominant creep mechanism. In this mechanism, vacancies diffuse from grain boundaries under tensile to boundaries in compression and metal atoms move in the opposite direction (see figure A2). These migrations result in the elongation of the grains (strain).



Figure A2: diffusion of vacancies (solid lines) and atom (dashed lines) due to applied stress [A1].

Diffusion can occur along the grain boundaries (D_{GB}), also known as Coble creep [A1,A4], or through the lattice of the grain (D_V), the so-called Nabarro – Herring (N-H) diffusion [A1,A5,A6]. For both diffusion processes, a ratio is found that links true strain rate \dot{e}_s to stress (σ), diffusion coefficient (D) and grain size (d, for diameter):

Coble:	$\dot{\epsilon}_S \div \frac{\sigma D_{GB}}{d^3}$,	(A8)	and
Nabarro – Herring:	$\dot{\epsilon}_S \div rac{\sigma D_V}{d^2}$		(A9)	

From both equations, the relation $\dot{\epsilon}_s \div \sigma$ is observed, which characterises viscous creep. Another observation is that Coble creep is more grain size sensitive than Nabarro – Herring creep: smaller grains mean more grain boundaries and Coble creep is more likely to occur. At lower temperatures, the D_{GB} is much larger than D_V , which causes Coble creep to be the dominant creep process.

A3.2 Intermediate stresses and high temperatures

At intermediate stress levels, it is shown that creep deformation is dominated by diffusion-controlled dislocation movement [A1,A3,A7,A8,A9]. These dislocations move through climb (edge dislocations) or cross-slip (partial screw dislocations) to avoid dislocation barriers. This behaviour has led to the so-called 'power-law creep':

$$\dot{\epsilon}_S \div \frac{D\sigma^n}{T}$$
 with n = 4 – 5 . (A10)

The diffusivity (*D*) depends exponentially on the temperature, resulting in an increase in strain rate with increasing temperature. At $T \ge 0.5 T_m$, dislocation creep is often the dominant creep mechanism [A1]. Two important observations for the power-law creep are that dislocation creep is independent on grain size and that dislocation creep depends on the shear modulus (*G*) through equation A7.

A3.3 High stresses and high temperatures

The most accepted reasoning for the exponential creep at high stress levels is an accelerated diffusion caused by an excess vacancy concentration due to interactions between dislocations [A1,A9]. Creep in this region is difficult to predict and should be avoided for long-life purposes for that specific reason.

A3.4 Grain boundary sliding

Grain boundary sliding (GBS) is a deformation mechanism accommodated by diffusional flow, which depends on stress direction, temperature, and grain boundary morphology [A1]. GBS is essential for grains to elongate into the stress direction and to keep the adherence between the grains. Concluded from these observations is that a material is more creep resistant when the number of grain boundaries is kept as small as possible. This implies that coarse-grained materials are less susceptible to grain boundary sliding, and thus creep, than fine-grained materials.

A4. Deformation – mechanism map

Deformation maps in stress – temperature space have been developed, showing the rate-controlling creep mechanisms, high-temperature deformation mechanisms and pure glide [A1,A10]. These maps are constructed for one average grain size. The normalised shear stress is the shear stress divided by the shear modulus. The deformation map indicates the dominant creep mechanism clearly as function of both stress and temperature. Strain-rate curves are plotted into the map for creep relation which relate the strain rate to the combined stress and temperature. An example for nickel is shown in figure A3, left figure. Here, at higher stresses, dislocation creep is distinguished between low-temperature (LT) creep due to climb through the dislocation core and high-temperature (HT) creep, which is caused by climb through the lattice. For smaller grains as shown for pure nickel with grain diameter of 10 μ m in the right figure of figure A3, the influence of the grain boundaries becomes clear. The boundary diffusion (Coble creep) region expands along the temperature axis of the deformation map at the expense of bulk/lattice diffusion (N-H creep). Also, diffusional creep is dominant at higher stresses in relation to dislocation creep at a fixed temperature. Thus, diffusional creep is significantly more dominant in materials with more grain boundaries. The larger influence of diffusional creep has shifted the iso-strain-rate curves to lower stresses and temperatures. So, fine-grained materials are more susceptible to creep than coarse-grained materials.



Figure A3: Stress-deformation map for 1 mm (left) and for 10 μ m (right) grain-size nickel [A10].

The stress-deformation map is a tool for determining the parameters for material processing, material testing and fields of application (Figure A4). The hot-working and hot-torsion regions are the regions at which the material is being processed to get the desired shape. Examples are forging and hot-rolling. The adjacent strain rate is especially due to the mechanical strain generated by hot-working.



Figure A4: Stress-deformation map for 100 μ m grain-size nickel, showing fields of processing and application of the material [A10].

Creep tests are performed with those parameters to come to a compromise between the insurance that steady-state creep strain rate is established (i.e. creep regime II) and time consumption for performing the creep test is reasonable. It should be considered that the application circumstances of the material are usually different than the lab test environment.

A creep strain is maximum 1%/year generally accepted for engineering design [A1,A10]. This is equivalent to a strain rate of approximately $3 \times 10^{-10} \text{ s}^{-1}$. This is well within the elasticity domain partly due to the safety factor used in designing. Choosing a temperature and a stress underneath the isostrain-rate curve of $3 \times 10^{-10} \text{ s}^{-1}$ will minimise the effects of creep on the material and the lifetime expectancy can be guaranteed with a high degree of confidence.

A5. Creep testing

Tests for determining creep cannot have the duration of the actual service life of the installation from time and cost perspective. For that reason, extrapolation of the data evolving from the (shortened) creep tests is essential. Many methods for extrapolation of the data have been developed, but the Larson-Miller parameter and the Shelby-Dorn parameter are the most widely used.

A5.1 Larson – Miller Parameter

The Larson – Miller parameter (LMP) is a combination of the Monkman-Grant relation (equation A2) and the Zener-Hollomon parameter (equation A3) resulting in:

$$LMP = \frac{\Delta H}{2.3R} = T(C_{\sigma} + \log t_R) \qquad , \tag{A11}$$

in which C_{σ} is a stress-dependent constant.

In this relation is assumed that the activation energy (ΔH) is independent of temperature and stress and thus a constant. For that reason, each material is believed to have a certain *LMP* for a constant applied stress [A12], which makes from C_{σ} a constant. For C_{σ} often the number 20 is taken on basis of experience, but it can be calculated as well from two sets of $t_R - T$ data at a certain stress [A1]:

$$C_{\sigma} = \frac{T_2 \log(t_R)_2 - T_1 \log(t_R)_1}{T_1 - T_2} , \qquad (A12)$$

or graphically from the $t_R - 1/T$ graph [A1]:

$$C_{\sigma} = -\log t_R \quad \text{for} \quad \frac{1}{T} \to 0 \qquad .$$
 (A13)

After C_{σ} is determined or chosen and the time to rupture (t_R) is determined for a certain temperature, the *LMP* can be calculated. From this *LMP* a prediction can be made from the time to rupture at a certain stress and temperature. Creep mechanisms cannot be determined from the *LMP*. The *LMP* is used to indicate the creep resistance. Alloy A is more resistant than alloy B in case [A1]:

• $LMP_A > LMP_B$ and $C_{\sigma,A} = C_{\sigma,B}$ or • $LMP_A = LMP_B$ and $C_{\sigma,A} < C_{\sigma,B}$

A5.2 Sherby – Dorn Parameter

The Sherby – Dorn parameter (SDP) is like the *LMP*, but uses a rupture time that is compensated for the temperature (θ). The advantage is that θ is independent from the temperature and this results in the curves for the different stresses being straight parallel lines in a ln(t_R) – 1/t graph. At the moment, it is difficult to determine an appropriate parameter for extrapolation and predictions using this parameter cannot be made confidently. Standardisation process for SDP is in development.

A6. Nickel alloys and creep

Nickel-based alloys (NA) are often chosen for high temperature purposes, because of their good creep resistance. Besides grain size, alloying of the material has a significant influence on the creep resistance [A1]. In table A1 the effect of the alloying elements in a NA are shown:

Alloying elements	W	Mo	Ti	Cr	Nb	Та	V	Со	Al	В	Zr	Hf
Effect												
Decreased self-diffusion due to high T _m	х	х										
Solid solution strengthening	х	x	х									
Solid solution strengthening due to high solubility				x								
Grain boundary carbides: block migration and sliding	х	x	x	x	x	x	x					
Stabilising of Ni ₃ (Al,X) precipitates (X = Ti, Nb, Ta)								х				
Surface stabilisers due to oxide formation				x					x	x	x	х

Table A1: alloying elements in a nickel-based alloy and their effects.

The effect of solid-solution strengthening is that it hinders dislocation movement and increases the stiffness of the material. Stable carbides on the grain boundaries restrict grain boundary migration and sliding. Ni₃(Al,X) precipitates, a.k.a. γ' , are formed in the nickel (γ) matrix and these precipitates have some unique characteristics. First, due to the lenticular shape of the γ' precipitates, dislocations have great difficulties to climb around the γ' precipitates [A1,A13]. Second, γ' precipitates are well-ordered and therefore retard dislocation movement through the precipitates, since that will disrupt the order [A1]. Finally, the γ' precipitates have the unusual behaviour to increase the strength of the NA by factor three to six with increasing temperature from ambient to ca. 700 °C [A1,A14]. The reasoning for this unusual strength increase is still under investigation. The effect of the surface stabilisers is reduction of oxidation rate (Cr and Al), hot corrosion (Cr) and oxide spalling (Al), and an improved hot strength, hot ductility and rupture life (B, Zr, Hf) [A1,A15].

Also, the heat treatment after hot-working and other processes, like welding, is important. Creep resistance of the NA is improved by solution treatment (annealing) to increase the grain size with subsequent quenching and ageing to obtain a structure with an optimal γ' precipitate distribution.

A7. Fracture mechanics of creep

Above the 'equi-cohesive temperature' the fracture path in a material changes from transgranular to intergranular because grain boundaries (GB) weaken faster than grains with increasing temperature [A1]. Intergranular fracture is a combination of grain boundary sliding (GBS) and cavitation (i.e. forming of micro-voids) on the grain boundaries.

GBS is especially occurring at temperatures larger than 0.4 T_h , and is controlled by dislocation creep at the lower temperatures and higher stresses, while GBS is caused by diffusional creep at the high temperatures. Regarding the creep rate, GBS is considered a viscous process ($\dot{\epsilon} \div \sigma$), while dislocation creep has a power-law dependence ($\dot{\epsilon} \div \sigma^{4...5}$). So, GBS involves a significant portion of the creep rate at low strain rates/stresses, while at high strain rates/stresses, GBS has a smaller part in the creep rate and deformation within the grains is the most dominant creep mechanism. Next to the presence of grain boundary particles (carbides), GBS is affected by the presence of solutes and the amount of grain mis-orientation [A1]. Grain boundary (GB) cavitation is commonly caused by de-cohesion between grain boundary particles and the matrix and it is associated with stress concentrations or structural irregularities, like triple points, GB ledges and hard particles [A1,A16]. Therefore, it is important to control the extent of impurity segregation on the GBs. The stability of a cavity depends on its radius [A1,A17,A18]:

$$r_c = \frac{2\gamma_s}{\sigma} \qquad , \qquad (A14)$$

in which:

 r_c = critical radius of the cavity

 γ_s = surface tension of the GB

 σ = tensile stress perpendicular to the GB

When $r < r_c$, the cavity surface sinters and closes up. In case $r \ge r_c$, the cavity is stable and continues to grow, resulting in coalescence and finally into intergranular fracture. Also, interstitials atoms have a more stabilising effect on cavities than substitutional atoms due to their size and a smaller grain boundary diffusion rate (D_{GB}) reduces the growth of the cavity. The latter is accomplished by hafnium (Hf) and borium (B) segregation to the GBs, which reduces D_{GB} by an order of magnitude [A1,A19]. Since the r_c depends on the orientation of the tensile stress, it is evident that GBs that are orientated perpendicular to the tensile stress are the most susceptible to cavity growth.

Initiation of intergranular creep fracture can occur in two ways. At lower temperatures and high stresses, the stress at triple points is high due to geometrical incompatibilities after GBS. This causes wedge cracks to initiate at the triple points. It is important to note that compressive stresses do not cause damage, only tensile stresses do. At high temperatures and relatively low stresses, diffusion along the GBs is significant and cavities can grow, resulting in void formation at the GBs, especially on those GBs that are orientated perpendicular to the load.

A8. Literature for Appendix A

- [A1] R. W. Hertzberg, *Deformation and fracture mechanics of engineering materials*, 4th edition, 1995, Published by: Wiley
- [A2] F. C. Monkman and N. J. Grant, Proc. ASTM 56, 1956.
- [A3] A. K. Mukherjee, J. E. Bird, J. E. Dorn, *Trans. ASM 62*, 1969.
- [A4] R. L. Coble, Journal of Applied Physics, v. 34, 1963.
- [A5] F. R. N. Nabarro, *Report of a Conference on the Strength of Solids,* Physical Society, London, 1948.
- [A6] C. Herring, Journal of Applied Physics, v. 21, 1950.
- [A7] J. E. Bird, A. K. Mukherjee, J. E. Dorn, Quantitative Relation between Properties and Microstructure, Israel Universities Press, Haifa, Israel, 1969.
- [A8] J. Weertman, *Trans. ASM v. 61,* 1968.
- [A9] J. Weertman, Journal of Applied Physics, v. 28, 1957.
- [A10] M. F. Ashby, *The Microstructure and Design of Alloys,* Proceedings, Third International Conference on Strength of Metals and Alloys, Vol. 2, Cambridge, England, 1973.
- [A11] H. J. Frost and M. F. Ashby, *Deformation-Mechanism Maps, The Plasticity and Creep of Metals and Ceramics*, Pergamon Press, 1982.
- [A12] F. R. Larson and J. Miller, Trans. ASME 74, 1952
- [A13] D. D. Pearson, F. D. Lemkey, and B. H. Kear, Superalloys 1980, ASM, Metals Park, OH, 1980
- [A14] R. R. Jensen and J. K. Tien, *Metallurgical Treatises*, J. K. Tien and J. F. Elliott, Eds., AIME, Warrendale, PA, 1981.
- [A15] R. F. Decker and J. W. Freeman, *Trans. AIME 218,* 1961.
- [A16] M. H. Yoo and H. Trinkhaus, *Metall. Trans. 14A*, 1983.
- [A17] E. D. Hondros and M. P. Seah, *Metall. Trans. 8A*, 1977.
- [A18] M. P. Seah, Phil Trans. Roy. Soc. London, 1980.
- [A19] J. H. Schneibel, C.L. White and R. A. Padgett, Proc. 6th Int. Conf. Strength of Metals and Alloys, ICSMA6, R. C. Gifkins, Ed., Pergamon Press, Oxford, UK, 1982.

Appendix B – High temperature fatigue

Fatigue effects are most influential at stress concentrations. These stress concentrations can be caused by design (geometry) or by flaws inside the material itself. At high temperatures, the material flaws are usually wedge cracks or voids/cavities caused by creep mechanisms. Other fatigue sites are due to or accelerated by oxidation and corrosion.

B1. Elastic and elastic-plastic behaviour

During elastic cycling, the material shows macroscopically a reversible strain behaviour. However, microscopically the material might have suffered some plasticity. This plasticity has evolved around stress concentrations induced by material flaws causing a local exceeding of the yield stress. This micro-scale plasticity in the material increases with each cycle resulting in high-cycle fatigue.

Elastic-plastic cycling leads to loading of the whole material above the yield stress. The elastic deformation is reversed completely during the cycle, but the plastic deformation is not, which may lead to fatigue failure after not that many cycles: low-cycle fatigue.

Due to cycling, the material can harden (i.e. due to work-hardening) or soften (i.e. due to recovery), which can have large influences on fatigue life. Cyclic hardening makes the material stronger resulting in less plastic strain development during a cycle, hence the material is more fatigue resistent. Cycling softening, a.k.a. the Bausinger effect, works opposite, so more plastic strain per cycle. For that reason, softening mechanisms are to be avoided in constructions exposed to cyclic loading. A basic rule for cyclic behaviour:

$\frac{\sigma_{UTS}}{\sigma_{YS}} > 1.4$	\rightarrow	Hardening
$1.2 < \frac{\sigma_{UTS}}{\sigma_{YS}} < 1.4$	\rightarrow	No or little effect
$\frac{\sigma_{UTS}}{\sigma_{YS}} < 1.2$	\rightarrow	Softening

B2. Fatigue lifetime

The lifetime of fatigue can be divided into two phases: the crack initiation phase and the crack propagation phase. Crack initiation usually occurs along slip planes, even at low mean stresses. An initiating crack grows slower in case the grain size is smaller, since grain boundaries are barriers for crack development. When an initiating crack becomes larger than the size of a grain, the crack is referred to as a propagating crack. Crack propagation moves in a direction perpendicular to the main loading axis, and failure occurs when a critical crack size is reached. The higher the mean stress, the faster the crack propagation. Large grain sizes lead to faster growth rates for a propagating fatigue crack. The lifetime of the material is predicted by elastic-plastic or linear-elastic fracture mechanics.

B3. High temperature fatigue characteristics

Fatigue at high temperatures shows different behaviour than fatigue at lower temperatures. First, the fatigue at high temperatures shows more low-cycle behaviour due to a lower yield strength. Second, the crack propagation transfers from transcrystalline to intercrystalline, since at lower temperatures the grain boundary is stronger than the grain, however generally that is vice versa at high temperatures (e.g. low melting phases). Last difference is the influence of creep on fatigue life at high temperatures. The introduction of creep makes the grain size important, since small grains make the material more fatigue resistant, while large grains lead to a better creep resistance. The maximum lifetime is obtained for an optimal grain size, which depends on the respective contributions of fatigue and creep.

B4. Nucleation of a high temperature fatigue crack

More mechanisms can cause the nucleation of a crack besides stress concentrations induced by design. Some of these mechanisms originate inside the material and others at the free surface.

B4.1 Cyclic slip

The first inner material mechanism is cyclic slip. Though cyclic slip is also active at low temperatures, the temperature enhances the cyclic slip process due to these reasons [B1]:

- Stacking fault energy increases with increasing temperature. This thermally activates slip and climb
- Larger oxidation and/or gas diffusion at higher temperatures enlarges the kinematic irreversibility of cyclic slip
- Stability of the microstructure changes at high temperature. The metallurgical changes can be either beneficial or detrimental for the resistance against fatigue.

B4.2 Sites caused by creep

Cracks in the inner material can nucleate on flaws induced by creep like wedge cracks, that are caused by grain boundary sliding (GBS), and cavities on the grain boundaries (GB) initiated by diffusion. Both creep mechanisms have been described in appendix A. GBS is more active and damaging in the inphase (i.e. tension stress due to thermal heating) cyclic loading than in the out-of-phase (i.e. compression stress due to thermal heating) cyclic loading, which on its turn is more destructive than isothermal mechanical fatigue cycling. Also, the shape of the stress cycle affects the cavitation content strongly [B1]. The cycle with slow tension loading and fast compression loading (i.e. longer stress holding time) results in a greater tendency of cavitation forming. The reason is that cavities can only develop in tensile mode and close up during compression.

B4.3 Inclusions and precipitates

De-cohesion of inclusions/second-phase particles from the surrounding matrix inside the grain of a microstructure may occur due to a difference in thermal expansion. This effect causes intergranular cavitation. Also, precipitates on the GBs are nucleation sites for voids. Cavitation and voids are stress raisers, which result in more cyclic irreversibility and for that reason they are sites for fatigue crack initiation.

B4.4 Oxidation and corrosion

Though it does not cause fatigue initiation, the presence of an oxidising or corrosive environment can influence the fatigue initiation mechanisms, especially at high temperatures [B1]. First, the sintering or closure of cavities might be prevented due to the presence of such an environment. Second, in case a GB at the material surface is not covered by a protective oxide layer, oxygen gas or other elements causing embrittlement may diffuse along this GB and react with GB precipitates. Cavities nucleate around the GB precipitates under the influence of an applied stress. Also, residual gases can form that increase the internal gas pressure, which might initiate cavities and accelerate their growth. Third, the surface diffusion and slip step oxidation combination enlarges the kinetic irreversibility of cyclic slip. This causes fatigue crack nucleation without the presence of other mechanisms. This mechanism is also known as the Fujita mechanism [B1]. Fourth, a GB at a free surface is usually a preferred oxidation site resulting in the generation of a microscopic notch (i.e. a stress concentration). Finally, the fifth mechanism is the break-down of the protective (oxide) layer due to cyclic loads. The absence of a protective layer causes oxidation and/or chemical attack, which are both accelerated by temperature. At high temperatures, the deterioration of fatigue resistance of many materials is for a large part caused by environmental interactions [B1].

B5. Model for life prediction

To predict the total cyclic life of materials at higher temperatures, quite some models have been developed. These models can be divided into three groups, which are the frequency-modified Coffin-Manson equation, damage accumulation models and strain range partitioning.

B5.1 Frequency-modified Coffin-Manson equation

The strain range-life relationship can be modified incorporating frequency to account for the cyclic frequency effect at high temperatures [B1]. The frequency-modified Coffin-Manson equation is:

$$\Delta \epsilon_p = c_f \left(N_f \, v_c^{k-1} \right)^{-\beta} \qquad , \qquad (B1)$$

in which $\Delta \epsilon_p$ is the plastic strain amplitude, N_f is the number of cycles until failure, and υ_c is the cycling frequency. The constant c_f and the exponents k and β are specific to the temperaturematerial-environment. These latter parameters are usually determined empirical. To predict creepfatigue life, understanding of stress-strain hysteresis loops is required.

B5.2 Damage accumulation models

In this model, mechanical fatigue and creep superpose linearly the accumulated damage. The fatigue part is described with the Palmgren-Miner rule by the expression $\sum \frac{n_i}{N_{f,i}}$ in which n_i is the number of fatigue cycles at stress amplitude $\Delta \sigma_i$ and the $N_{f,i}$ is the number of cycles to failure at this stress amplitude. The sum is taken over the total number of stress blocks with different amplitudes. The creep part is determined similar: t_j is the exposure time under an average stress σ_j and t_{Rj} is the time to rupture at that stress level. The total amount of creep is the sum, so $\sum \frac{t_j}{t_{Rj}}$. The accumulated damage (d_{tot}) is the sum of the fatigue and creep parts [B1]:

$$\sum \frac{n_i}{N_{f,i}} + \sum \frac{t_j}{t_{Rj}} = d_{tot} \tag{B2}$$

Three considerations have to be taken into account when applying this model [B1]. First, stresses continue to relax under fixed strain dwell periods. Second, cyclic hardening and softening leads to different levels of stress relaxation during continued fatigue straining. Finally, stress rupture data is usually derived from monotonic load tests and are generally not representative for in-service creep-fatigue behaviour. In case the linear damage accumulation model is used, this is done in conjunction with strain-controlled fatigue tests with dwell periods under load [B1].

B5.3 Strain range partitioning

The strain range partitioning models divide the overall fatigue damage into mechanical and timedependent components of the strain range. Time-independent mechanical fatigue is caused by cyclic slip, while GBS and GB cavitation are the mechanisms leading to creep-fatigue damage (i.e. timedependent damage). This model considers the reversal of strains related with fatigue and creep. The reversed inelastic strain is divided into four components, which are shown in figure B1.



Figure B1: The four components of inelastic strain range that are basis for strain range participation. E, P and C indicate respectively elastic, plastic and creep deformation [B1].

 $\Delta \epsilon_{pp}$ is tensile plastic deformation reversed by compressive plastic deformation, $\Delta \epsilon_{cp}$ is tensile creep reversed by compressive plastic deformation, $\Delta \epsilon_{pc}$ is tensile plastic deformation reversed by compressive creep, and $\Delta \epsilon_{cc}$ is tensile creep reversed by compressive creep. The total fatigue life (N_f) is determined by the sum of the four components of inelastic strain through the (unmodified) Coffin-Manson equation [B1]:

$$\frac{1}{N_f} = \frac{1}{N_{pp}} + \frac{1}{N_{cp}} + \frac{1}{N_{pc}} + \frac{1}{N_{cc}} \qquad , \tag{B3}$$

in which the subscripts of these cycle portions denote the same modes as those of the strain ranges. Many empirical modifications to this model are available to deal with the effects of slow-fast fatigue loading and with the influence of creep-fatigue dwell periods into the life calculation [B1].

B5.4 Crack growth rate

The severity of a crack can be determined by the crack growth rate. Since often only the fatigue crack data is known or only the creep crack data, it is convenient to combine those data. The growth rate is then determined by the sum of the fatigue crack growth rate and the creep crack growth rate, which are determined by fracture mechanics models:

$$\frac{da}{dN} = \left(\frac{da}{dN}\right)_{fatigue} + \left(\frac{da}{dN}\right)_{creep}$$
, (B4)

with:

$$\left(\frac{da}{dN}\right)_{fatigue} = C_f (\Delta K)^{m_f}$$
 , (B5) and

$$\left(\frac{da}{dN}\right)_{creep} = \int_0^{1/\nu_c} \frac{da}{dt} dt \quad and \quad \frac{da}{dt} = C_c (C^*)^{m_c} \qquad . \tag{B6}$$

Here $\frac{da}{dN}$ is the crack growth per cycle and $\frac{da}{dt}$ is the crack growth per time unit. C_f and C_c are material constants for respectively fatigue and creep mode. C^* is a time-dependent variant describing creep crack growth, ΔK is change in stress intensity factor and m_c and m_f are material exponents for respectively fatigue and creep mode. The choice of the most suitable data available is essential for a sufficient accurate determination/estimation of the growth rate. An appropriate safety margin is required to meet any erroneous assumptions when choosing the data.

B6. Thermal fatigue and thermo-mechanical fatigue

Roughly high temperature fatigue can be divided in thermal fatigue (TF) and in thermo-mechanical fatigue (TMF) [B2].

TF stresses are caused by temperature gradients and by thermal expansion differences of individual grains/phases in the material [B2]. The temperature gradients depend on temperature fluctuations (i.e. temperature depending on time (T(t)) and the heat conductivity of the material). A variety in isotropy of adherent grains or dissimilar phases (like matrix and carbides) cause differences in thermal expansion, which results in fatigue loading with a cyclic varying temperature.

TMF stresses are the result of mechanical strains developed by thermal strains combined with external mechanical constraints [B2]. Thermal strain (ε_t) is calculated from the thermal expansion coefficient (α) and the temperature difference between temperature T_1 and temperature T_2 .

$$\varepsilon_t = \alpha (T_2 - T_1) \tag{B7}$$

The total strain (ε_{total}) is the surplus of the thermal strain and the mechanical strain (ε_{mech}).

In case of total constraint (i.e. ε_{total} is zero), the ε_t and the ε_{mech} are equal, but have an opposite sign. With partial constraint, the ε_t is only partly counteracted by the ε_{mech} leading to some thermal expansion or contraction. The ε_{mec} is not present in a situation without mechanical constraint and the ε_{total} solely depends on ε_t , which means that no reaction stresses develop in this situation.

B6.1 Shakedown

The balance between ε_t and ε_{mech} under total or partial constraint lead to a behaviour known as shakedown. In the simplified model with total constraint, as shown in figure B2, it is assumed that the Young's modulus (*E*) is independent of the temperature, that the system is free from hardening effects, and that the σ_y is equal in tension and in compression load. In the first cycle, the material is heated from temperature T_0 , causing the material to expand (ε_t). Since the material is constraint, the ε_{mech} increases with the same amount as ε_t , but then reciprocally, resulting in a compressive stress in the material. First, the ε_{mech} is elastic strain until the σ_{ys} (i.e. $\varepsilon_{mech} = \frac{-\sigma_y}{E}$) is reached. In figure B2, this is shown with the line from the origin to point A and the adherent temperature at point A is set as T_1 . Upon further heating to T_2 at point B in figure B2, the ε_{mech} continues in the plastic region with the amount of $-\alpha(T_2 - T_1)$. From point B onwards, there are two scenarios.

In the first scenario, cooling starts and the elastic strain is reversed until temperature T_0 is obtained at point C in figure B2. The compressive stress is reduced with $-\sigma_y + \alpha \cdot E(T_2 - T_0)$ and a residual tensile stress has appeared. Since the residual tensile stress is smaller than the σ_y , there is no plastic straining in tensile mode. Thermal cycling between T_0 and T_2 results in elastic straining only between points B and C. This behaviour is known as elastic shakedown [B2]. Due to the small amount of initial plastic straining during the first cycle and no further plastic straining during the subsequent cycles, the fatigue related to elastic shakedown is referred to as high-cycle fatigue [B2].



Figure B2: Shakedown, both elastic and plastic [B2].

The second scenario involves further heating from T_2 (point B) to T_3 (point D), which causes more plastic ε_{mech} . At T_3 , the ε_t has caused the plastic ε_{mech} to become larger than the elastic ε_{mech} . Upon cooling to T_0 , the compressive elastic ε_{mech} is reversed taking the stress from compressive to tensile mode until the tensile σ_y is obtained (i.e. elastic $\varepsilon_{mech} = \frac{2\sigma_y}{E}$). At that point, not all the recovered ε_t has been compensated for by ε_{mech} , so the remainder of the ε_{mech} is plastically strained by tensile stress until point E is reached (where $\varepsilon_t = \varepsilon_{mech}$ at T_0). Cyclic loading between T_0 and T_3 will always cause subsequent plastic straining, which is shown in figure B2 by the hysteresis loop [B2]. This process is called plastic shakedown and the additional damage with each cycle leads to low-cycle fatigue [B2].

B6.2 Ratchetting

Ratchetting is the fatigue process in which plastic strains accumulate during each subsequent cycle and therewith continuously elongating the wall exposed to thermal stress with each cycle. Often, the two-bar model of Morrow (figure B3) is used to illustrate the ratchetting mechanism. In this model, bar 1 is exposed to a cyclic temperature, while bar 2 is maintained at a constant temperature.



Figure B3: Morrow's two bar model [B2].

The model is subject to a constant load (P_n) which can be rewritten as:

$$P_n = \sigma_1 A_1 + \sigma_2 A_2 \tag{B8}$$

In which σ_1 and σ_2 are the stresses in respective bar 1 and bar 2 and A_1 and A_2 are the cross-section areas of respectively bar 1 and bar 2. The expansion (δ) in both bars is equal ($\delta_1 = \delta_2$), which can be expressed as:

$$\varepsilon_1 L_1 = \varepsilon_2 L_2 \tag{B9}$$

In which L_1 and L_2 are the lengths of respectively bar 1 and bar 2. The strain in bar 1 (ε_1) consists out of an elastic component (ε_1^{el}), a plastic component (ε_1^{pl}), and a thermal component (ε_1^{th}). The strain in bar 2 (ε_2) does not have the thermal exponent, since it is not subject to thermal cycling.

$$\varepsilon_1 = \varepsilon_1^{el} + \varepsilon_1^{pl} + \varepsilon_1^{th} \tag{B10}$$

$$\varepsilon_2 = \varepsilon_2^{el} + \varepsilon_2^{pl} \tag{B11}$$

With ratchetting one bar yields during the heating portion of the thermal cycle, while the other bar yields in the same direction during the cooling part of the thermal cycle. An accumulation of strains is the result, which causes a combined fatigue and ductility exhaustion mechanism [B2].

B6.3 Stress – strain operating regimes

To determine which of the four stress-strain regimes can be expected, a diagram is constructed as shown in figure B4 (modified from [B2]). Along the vertical axis is displayed the ratio between the thermal stress (i.e. combination of Young's modulus (E), the thermal expansion coefficient (α), the temperature cycle (ΔT)) and the yield strength (σ_y). On the horizontal axis, the ratio between the applied load (P_n) and the limit load (i.e. maximum load; P_l) is displayed. Commonly, P_l is equal to σ_y . Both σ_y and P_l are determined at reference temperature T_0 , which is often the service temperature.



Figure B4: the four stress – temperature regimes [B2, modified]; E.S. is elastic shakedown, P.S. is plastic shakedown.

B7. In-phase and out-of-phase thermo-mechanical fatigue

A phase relation exists between ε_{mech} and the temperature. In-phase (IP) TMF means that the peak strain coincides with the maximum temperature, while in out-of-phase (OP) TMF the peak strain is obtained at the minimum temperature. This is illustrated in figure B5.

It is important to note that maximum stress does not occur at the maximum temperature. Also, the situation that during IP TMF the stress is not maximum at T_{max} and maximum ε_{mech} is due to softening mechanisms occurring with increasing temperatures (i.e. σ_y is lowered). The same applies to OP TMF, but than with minimum stress (compression) instead of maximum stress. During cooling the σ_y can also increase due to (cyclic) hardening effects. Hardening is beneficial, since with the increase in σ_y , the adjacent ε_{pl} is reduced (i.e. less permanent damage). In IP TMF more plastic strain evolves in tension mode, while in OP TMF the greater part of the plastic strain occurs in the compression mode. In case the ε_{mech} is zero, the mean stress in IP TMF is negative and the mean stress for OP TMF is positive.



Figure B5: left figure: variation in ε_{me} /T in IF (isothermal fatigue), TMF IP, and TMF OP; Right figure: TMP IP and OP σ - ε response [B2].

For designing purposes, the plastic strain range of a thermo-mechanical cycle can be determined:

$$\Delta \varepsilon_{pl} = \Delta \varepsilon_{mech} - \left(\frac{\sigma_{T_{max}}}{E_{T_{max}}} + \frac{\sigma_{T_{min}}}{E_{T_{min}}}\right) \tag{B12}$$

This is an approximating which underestimates the plastic strain range slightly [B2]. More accurate equations do exist, but these are for the greater part on the same basis as equation B12.

B8. TMF of high-temperature nickel-base alloys

Nickel-base (super)alloys (NA) have a TMF operating rate up to $T_h = \frac{T_{max}}{T_{melt}} = 0.8$, in which T_h is the homologous temperature (i.e. temperature ratio), T_{max} is the maximum temperature in the thermal cycle and T_{melt} is the melting temperature, all in degrees Kelvin [B2]. The effects of strain and temperature on hardening and softening is not yet understood, but for Alloy 617 it has been observed that significant hardening occurred at a T_{max} of 750 – 850 °C and that at $T_{max} > 950$ °C, the strain response is stable [B2, B3]. A greater part of the hardening is caused by dislocation pile-ups at the interfaces between precipitates (γ') and matrix. Also for Alloy 617, TMP IP damage is more significant than TMF OP damage when fluctuating the temperature between $T_{max} = 600$ °C and $T_{max} = 850 - 1050$ °C [B2, B3]. This is expected, since commonly tensile stress causes more damage to the microstructure than compressive stress.

B8.1 Effects of strain rate, time and frequency

Strain rate affects both fatigue life and cyclic stress-strain response, though the results are not straight-forward. Roughly there is a change of the strain-life curve from IP to OP, and vice versa, at approximately $\Delta \varepsilon_{pl} = 0.0045$. Lifetime in IP is shorter for high $\Delta \varepsilon_{pl}$ because of excessive creep crack growth caused by large tensile strains at high temperatures. At low $\Delta \varepsilon_{pl}$, OP lifetime is shorter due to abundant oxidation (long exposure times) resulting in brittle fracture induced by large compressive strains. A decrease in strain rate, holding times and/or loading frequency results in a stress range reduction and in shorter cyclic life. This is caused by an increase of damage by creep and environment.

B8.2 Microstructural changes during TMF

During TMF both thermal aging and dynamic strain aging occur. The thermal aging effect is shown as dislocations networks formation around the γ' precipitates and by coarsening of the γ' precipitates (i.e. thermal recovery). Dynamic strain ageing depends on both strain rate and temperature and it leads to cyclic hardening or softening in solute-hardened NAs. Usually hardening occurs at lower temperatures because of the formation of precipitates and softening happens at high temperatures due to the coarsening of precipitates [B2]. Softening leads to a stress reduction and hence a reduction in creep.

On the grain boundaries (GB) damage may occur due to uneven strains during a cycle. These uneven strains can induce local ratchetting at voids and wedge cracks that are caused by creep. Also, carbides can precipitate at GBs, which makes the material susceptible to embrittlement [B2].

B8.3 Environmental effects

Environmental damage can affect crack initiation and crack propagation and has a detrimental effect on fatigue life as described earlier in this chapter. Oxidised samples with a stress raiser (notch) can have crack growth rates that are three orders of magnitude larger than un-oxidised samples without a stress raiser [B2].

At high temperatures, a protective oxide film forms on the outer surface separating the substrate from the environment. Due to stress development in the film during thermal cycling (i.e. different thermal expansion coefficient), the film suffers from spalling and cracking. The oxide film and the oxidation characteristics of NA depends on the alloy composition, temperature and the exposure time. In NA the oxide layer consists often out of Al_2O_3 or Cr_2O_3 layers or complexes (spinels) of those elements. The oxidised surfaces are usually associated with an adjacent zone depleted from γ' precipitates, which dissolve due to the loss of aluminium to the oxide, and from the solid-solution-strengthening element chromium, that has high affinity with oxygen as well [B2]. The loss of γ' precipitates and solidsolution-strengthening elements lower the fatigue resistance of the NA. It has been shown that applying a stress increases oxidation and alloy depletion [B2].

B8.4 Overview damage mechanisms

In the ASM handbook an overview is given of damage mechanisms in NAs, divided in IP and OP.

Material	TMF in-phase	TMF out-phase	Reference
Coating Ni-base superalloys		Fracture of coating upon cooling below its ductile brittle transition	<u>158</u>
Mar-M247 (uncoated polycrystalline)	Intergranular crack initiation and growth	Repeated oxide damage rafting of the γ ' structure different than IF or TMF IP	<u>105</u>
Mar-M247 (coated and uncoated)	Intergranular crack initiation and growth, internal crack initiation Rafting of the γ' structure different than IF or TMF OP	Fracture of coating upon cooling below its ductile brittle transition Rafting of the γ' structure different than IF or TMF IP	<u>107, 135</u>
Hastelloy X (solution strengthened Ni-base superalloy)	Strain rate dependent dynamic strain aging Precipitation hardening due to Cr-rich M ₂₃ C ₆	Strain rate dependent dynamic strain aging Precipitation hardening due to M ₂₃ C ₆	<u>103</u>
AM1 single crystal		Environment initiated damage in TMF differs from casting defect initiated damage in IF	<u>48</u>
IN 738 LC (coated and uncoated)		Lower lives for coated OP case because of coating fracture	118
CMSX-6 (single crystals)	Soft γ' matrix formation, cutting of particles at low temperatures, dislocation climb during high-temperature portion of the cycle		108

Figure B6: overview of damage mechanisms for IP and OP TMF in NAs [B2].

B9. Test methods to investigate TF and TMF

Over the years many test methods have been proposed and examined, especially for heating the material during the test. In figure B7 these methods are summarised including their advantages and drawbacks. Also listed are the materials used in these methods. From the overview of the methods is shown that quite some metallurgical, geometrical and financial drawbacks accompany the high-temperature fatigue testing.

Type of test	Heating method	Advantages	Disadvantages	Materials studied
TF	Immersion in hot and cold oil bath	Simplicity of the experiment	Transient stress strain could be present and should be calculated	Noncubic crystals, including tin, zinc, cadmium
TMF	Direct resistance	Rapid heating; allows space to mount the extensometer and pyrometer for crack growth measurements	Electric isolation of grips; local heating of crack tips	Conductive materials, stainless steel
TMF and TF	Induction (10-450 kHz, 5-40 kW capacity)	Rapid heating; complex specimen geometries permitted; inert environment testing using bellows	Experience with coil design required; electric noise in the strain signal due to high-frequency magnetic fields; high cost of unit	Aluminum, copper, steels, nickel-base superalloys
TMF and TF	Quartz lamp (radiation)	Inexpensive; uniform temperature over different zones of the specimen	Shadow effects; slow cooling rates; enforced cooling needed	Nickel- and cobalt- base superalloys, metallic composites
TF	Fluidized bed	Good for screening TF resistance of materials	Stress-strain temperature transients must be calculated and surface oxidation removed	Nickel-base superalloys
TF	Burner heating; flame heating	Good for screening TF resistance of materials; surface hot corrosion damage representative of service	Stress-strain temperature transients must be calculated	Nickel-base superalloys, steels
TF	Thermal fatigue under bending	Under reversed bending one surface undergoes OP, the other undergoes IP	Stress-strain gradients must be calculated	Nimonic alloys
TF	Dynamometer (friction heating)	Very high temperatures on surface reached; representative of service	Oxides are wedged into cracks; friction characteristics change with time	0.5 to 0.7% C steels

Figure B7: TF and TMP test methods [B2].

B10. Literature for appendix B

- [B1] S. Suresh, *Fatigue of Materials, 2nd edition,* Massachusetts Institute of Technology, Cambridge University Press, 1998.
- [B2] ASM International, ASM Handbook Volume 19, Fatigue and Fracture, 2003.
- [B3] Y. Pan, K. Lang, D. Lohe, E. Macherauch, *Cyclic Deformation and Precipitation Behavior of NiCr22Co12Mo9*, Phys. Status Solidi (a), Volume 138, 1993.

Appendix C – Fitness-For-Service Assessment Procedure

Fitness-For-Service (FFS) assessment procedures have been developed to evaluate damage or flaws in pressurised components and these procedures have been described in the API 579-1/ASME FFS-1 (edition 2007). In this standard, the assessment of components in service in the creep range is dealing with the damage mechanisms at elevated temperatures. A guidance on the manner to deal with multiple damage mechanisms acting simultaneously is included in this assessment procedure. In the research at hand, the effects of (crack-like) flaws and welds are not considered.

C1. Setup of the assessment procedure

The FFS assessment is generally generated following eight steps [C1]:

- 1. Identification of damage/flaw mechanism.
- 2. Applicability and limitations of the FFS assessment procedure.
- 3. Data requirements (e.g. material properties, design data, service conditions).
- 4. Assessment techniques and acceptance criteria.
- 5. Remaining life evaluation.
- 6. Corrective actions (remediation) to stop/reverse damage inflicted.
- 7. In-service monitoring (in case remaining life or inspection interval cannot be determined).
- 8. Documentation for qualification of the component (data, calculated risks, considerations, etc).

Among the required data (step 3) for boiler components and piping components are:

- ASME Manufacturer's Data Report or other equivalent documentation/specifications.
- Fabrication drawings, line lists and/or isometric drawings with detailed sketches to allow the performance of Maximum Allowable Working Pressure (MAWP) and stress calculations.
- Design calculations.
- Inspection records of fabrication.
- Material test reports.
- Pressure-relieving device information.

Besides the required data, a detailed operational and maintenance manual should be written, which contains information like the actual operating envelope (i.e. pressure and temperature), allowable corrosion rate, repair procedures and inspection frequency.

The assessment (step 4) is divided into three levels, which are interrelated. Each assessment level is a balance between the complexity of the analysis, amount of evaluation information required, skill of personnel performing the assessment, and conservatism. Level 1 is the easiest and most conservative assessment level and level 3 is an assessment based on numerical techniques (e.g. finite element method (FEM)) or based on experimental techniques. Level 2 is in-between level 1 and level 3. Acceptance criteria (step 4) are based on the allowable stress, remaining stress factor (RSF), and failure assessment diagram (FAD). The RSF is used to approximate the load carrying capacity of a component after some service time and the FAD is used to evaluate crack-like flaws.

The FFS assessment procedures assume that all required information is known. In case some variables are not known or not accurately determined, conservative estimates are made to ensure a reasonable

safety margin. Several types of analyses can be used to investigate the input parameter influence on the FFS assessment procedure results:

- Sensitivity Analysis: determination of an independent variable influence on the safety factors.
- Probabilistic Analysis: evaluation of the safety margin dependence on the uncertainty of independent variables.
- Partial Safety Factors: combining the individual safety factors of the independent variables, which are probabilistic determined for calculating the uncertainty in the assessment.

The remaining life evaluation (step 5) is not meant as a precise approximation of the actual lifetime, but has to be performed to establish an in-service monitoring plan, an appropriate inspection interval, or the need for remediation. Remediation (step 6) is applied in case the uncertainty in certain circumstances becomes to large (e.g. acceptability of flaw size, adequate assessment cannot be performed or estimated remaining life is difficult to approximate), and for each FFS assessment procedure appropriate remediation methods have been established. In-service monitoring (step 7) may be used to assess future damage conditions, and this method is usually applied in case future damage rate cannot be estimated accurately or easily. All the previous steps shall be documented (step 8) to ensure traceability and mechanical integrity compliance.

C2. Assessment procedure for service in the creep range

In the creep range, the FFS assessment procedure needs an approximation of the remaining life. This needs to be provided for components with and without a crack-like flaw subject to steady state and cyclic operating conditions.

C2.1 Applicability and limitations

The assessment in the creep range is not required in case the equipment is designed according to a recognised code or standard, though FFS evaluation is for example needed in case [C1]:

- Deviations in operating temperature, pressure, and/or loading conditions that may result in creep and which are not accounted for in the original design.
- Metal loss in the component larger than originally designed for.
- Component weldments with significant different properties in weld metal, heat affected zone (HAZ) and base metal.
- Stress concentrations regions that are not in the original design.
- Discovery of a (crack-like) flaw.
- Discovery of a local damage that may result in localised creep strain accumulation.
- Fire damage.

C2.2 Assessment level

Level 1 assessment can only be used in case no cycling is present. Level 2 assessment is applicable for components subject to maximal 50 cycles of operation. Both levels are therefore unsuitable for the FFS assessment in creep-fatigue circumstances. Level 3 assessment is required for cyclic operation in the creep regime, for complicated loading conditions, for complicated geometries, and for the presence of (crack-like) flaws.

C2.3 Data requirements

For the level 3 assessment in the creep range, it is required to perform a stress analysis, which includes the effects of stress re-distribution during creep. It is allowed to use numerical analysis techniques (e.g. FEM). In the stress analysis, several factors need to be considered, like material creep response, loading conditions and geometry of the component. The analysis should also regard other material properties, such as thermal conductivity and transient effects, and environmental circumstances, such as process environment on local overheating and insulation effects of corrosion/oxidation products.

The material data required to perform a remaining life assessment is provided in the API 579-1/ASME FFS-1. First, there is the MPC Project Omega Data, which provide data in terms of a damage parameter and strain-rate parameter and can also be used to analyse creep buckling. Creep rupture data is the second, which gives material properties in terms of the Larson-Miller parameter. Both data regard minimum and average properties. In a level 3 assessment also other sources for data may be used.

C2.4 Damage Characterisation/Acceptance Criteria

Damage can evolve in several manners and the damage accumulation mode(s) need to be considered. Besides the material conforming to the material specification, the material condition after a certain time in service must be assessed. Aspects to consider are [C1]:

- Remainder of sound wall thickness.
- Existence of damage or flaws.
- Local variations in service conditions.
- Unusual loading (e.g. missing supports).
- Environmental interaction (e.g. oxidation, corrosion).
- Grain size and heat treatment conditions.
- Welding issues.

The assessment should contain a recommendation about the non-destructive test (NDT) methods for determining possible damage. Common NDT methods are magnetic particle testing (MT) and liquid penetrant testing (PT) for surface indications, X-ray testing (RT) and ultrasonic testing (UT) for volumetric indications, and replica testing to determine the microstructure at the surface of the material. Most of these methods cannot be used at high temperatures (usually max. 50 °C).

C2.5 Remaining life evaluation

Six assessment procedures are provided for in the creep range [C1]:

- 1. Creep rupture life: for components subject to steady state operation without crack-like flaws.
- 2. Creep-fatigue interaction: for components subject to cyclic operation without crack-like flaws.
- 3. Creep crack growth: for components containing crack-like flaws and subject to either steady state or cyclic operation.
- 4. Creep buckling: for components that may become structural instable due to a compressive stress field. This procedure can be applied for components with or without a crack-like flaw.
- 5. Creep-fatigue assessment of dissimilar weld joints.
- 6. Microstructural approaches, which are used to supplement other assessments

Since the wall thickness of heat-exchanger tubes is thin (i.e. 1.2mm), crack-like flaws are considered detrimental directly (i.e. total loss) and a further assessment is not required, hence assessment procedure number 3 is not investigated. Assessment procedure number 5 is not considered since dissimilar welds are not part of the research.

Besides the API 579-1/ASME FFS-1 other assessment procedures may be used as an alternative, like British Energy R-5, BS 7910, EPRI "Remaining-Life of Boiler Pressure Parts – Crack Growth Studies", WRC 440 "A Synthesis of the Fracture Assessment Methods Proposed in the French RCC-MR Code for High Temperature", or ASME Code, Section III, Subsection NH.

C2.5.1 Sensitivity Analysis

The assessment procedures do not submit any recommendations for in-service margins, but it is better to apply an in-service margin to the remaining life calculation rather than applying one to each independent variable. The remaining life in the creep range depends for example on material data, stress and temperature. Confidence in the assessment may be gained when the sensitivity analysis proves that small changes in the independent variables do not lead to severe remaining life reductions. Of course, more uncertainty in each of the independent variables results in a larger in-service margin.

C2.5.2 Creep rupture life

The assessment is based on stress and strain locally and through the wall thickness of the component, and on the associated operating time and temperature. If an inelastic analysis is used to evaluate the effects of creep, then a material model is required to compute the creep strains in the component as a function of stress, temperature, and accumulated creep strain or time. If the computed stresses exceed the yield strength of the material at temperature, plasticity should also be included in the material model. The assessment procedure provides a systematic approach for evaluating the creep damage for each operating cycle. The total creep damage is computed as the sum of the creep damages calculated for each cycle [C1].

- a. STEP 1 Determine a load history based on future planned operation. The load histogram should include all significant operating loads and events that are applied to the component. If there is cyclic operation, the load histogram should be divided into operating cycles. Define *M* as the total number of operating cycles.
- b. **STEP 2** For the current operating cycle m, determine the total cycle time, ${}^{m}t$, and divide the cycle into a number of time-increments, ${}^{n}t$. Define N as the total number of time-increments in operating cycle m. The time-increments used to model the operating cycle should be small enough to capture all significant variations in the operating cycle (including changes in the wall thickness due to corrosion or erosion).
- c. **STEP 3** Determine the assessment temperature, ${}^{n}T$, for the time-increment ${}^{n}t$.
- d. **STEP 4** Determine the stress components, ${}^{n}\sigma_{ij}$, for the time-increment ${}^{n}t$. The principal stresses are computed using a finite element (FE) analysis.
- e. **STEP 5** Determine if the component has adequate protection against plastic collapse.
 - 1) If the stress components are determined from an elastic analysis, determine the primary load reference stress (${}^{n}\sigma_{ref}^{P}$):

$${}^{n}\sigma_{ref}^{P} = \frac{{}^{n}P_{b} + \left({}^{n}P_{b}^{2} + 9 \cdot {}^{n}P_{l}^{2}\right)^{0.5}}{3} \qquad , \tag{C1}$$

with P_b is bending stress and P_l is local membrane stress. Check that the following criterion is satisfied:

$${}^{n}\sigma_{ref}^{P} \le 0.75 \cdot \sigma_{y} \tag{C2}$$

The value of the yield strength (σ_y) is evaluated at temperature ${}^{n}T$ at time-increment ${}^{n}t$. If the criterion is satisfied, proceed to STEP 6, otherwise, proceed to STEP 12.

- 2) If the stress components are based on an inelastic analysis that includes plasticity and creep, protection against plastic collapse can be determined by a limit load or plastic collapse solution for example in accordance with API 579-1/ASME FFS-1.
- f. **STEP 6** Determine the principal stresses, and the effective stress (σ_e), using Von Mises criterion for the time-increment ^{*n*}t.
- g. **STEP 7** Determine a remaining life at the stress level ${}^{n}\sigma_{e}$ and temperature ${}^{n}T$ for timeincrement ${}^{n}t$ by utilising creep rupture data for the material. The equations suggested by API 579-1/ASME FFS-1 are not covering nickel-based alloys. Therefore, the available creep data has been gathered and a trendline has been determined, and the polynomic equation of the trendline has been used.
- h. **STEP 8** Repeating STEPS 3 to 7 for each time-increment ${}^{n}t$ in the mth operating cycle to determine the time to rupture ${}^{n}t_{R}$ for each increment.
- i. STEP 9 Creep damage accumulation for all points in the mth cycle can be computed:

$${}^{m}D_{c} = \sum_{n=1}^{N} \frac{{}^{n}t}{{}^{n}t_{R}}$$
 , (C3)

- j. **STEP 10** Repeating steps 2 to 9 for each of the operating cycles defined in Step 1.
- k. **STEP 11** Total creep damage (D_c^{total}) calculation for all operation cycles:

$$D_c^{total} = \sum_{m=1}^{M} {}^m D_c \le D_c^{allow}$$
(C4)

- I. **STEP 12** Conclusion of creep accumulation prediction. The total allowable creep damage should be taken as $D_c^{total} = 0.80$, though an alternative value can be used if justified. Also to be considered:
 - 1) The criterion for protection against plastic collapse has to be satisfied for any point in the operating history. When this criterion is not met, action is needed (e.g. de-rating).
 - 2) In case the D_c^{total} is less than the allowable creep damage D_c^{allow} then the component is acceptable for continued operation. The remaining life for operation is determined as the time when $D_c^{total} = D_c^{allow}$.
 - 3) If the D_c^{total} is larger than the D_c^{total} , the life is limited to the time corresponding to $D_c^{total} = D_c^{allow}$. In this case actions like de-rating have to be taken.

Inelastic analyses

An inelastic analysis of the stress components is required in case both plasticity and creep are active. In addition to the creep damage criterion, the total accumulated inelastic strains should be limited to a value that does not affect the operability of the component. A suggested limit for accumulated strains is provided in API 579-1/ASME FFS-1, Annex B1, par. B3.3.

Fired heater and boiler tubes

Fired heater and boiler tubes can be considered a specific case according to the API 579-1/ASME FFS-1. Requirements for this specific case are an approximately constant pressure and a temperature that is approximately uniform around the circumference of the tube during operation. The latter is not the case in the solar boiler, and therefore the pipe equation cannot be used directly. FEM analysis has to take into account the temperature variation around the tube's circumference.

C2.5.3 Creep-Fatigue interaction

Creep-fatigue interaction is part of the fit-for-service assessment for service at high temperature in combination with cyclic loads. The combination of creep and fatigue damage is handled in the report in paragraph 2.8.3 and 3.4.

C2.5.4 Creep crack growth

The effects of a crack-like flaw can be determined using the results from a stress analysis. The assessment deals with local and through the wall thickness stress and strain. Accumulated creep strain (strain-hardening) or time to rupture must be considered in case an inelastic analysis is used to evaluate the creep effects.

Creep crack growth is not ragarded in this model. The tube wall thickness is 1.2 mm and the pressure difference between the interior and exterior of the tube is approximately 40 bar (4 MPa). These circumstances make that a crack-like flaw will affect the integrity of the component significantly and is therefore not acceptable.

C2.5.5 Creep buckling

An in-service margin for protection against buckling collapse must be satisfied to avoid buckling of components subject to a compressive stress. There are two kinds of buckling considered [C1]:

- 1. Time-independent buckling as evaluated in accordance with for example with API 579-1/ASME FFS-1, Annex B1, par. B1.4.
- 2. Time-dependent buckling: to protect against load-controlled creep buckling. Instability is not allowed to occur within the total operational time determined for the specified loading multiplied by an in-service margin of 1.5. For pure strain-controlled buckling an in-service margin of 1.0 may be used since the strain-controlled loads are reduced concurrently with resistance of the structure to buckling when creep is significant.
 - a. Strain-controlled buckling is characterised by the immediate reduction of strain induced load upon initiation of buckling. Load-controlled buckling is characterised by continued application of a load in the post buckling regime, which lead to failure (e.g. collapse of a tube under external pressure). In case the two buckling loads interact, the in-service margin of the load-controlled buckling shall be used. This is also valid for cased with significant elastic follow-up.
 - b. The critical buckling time is determined based on the effects of original fabrication tolerances, geometrical imperfections and other flaw types.

- c. Calculations (including isochronous stress-strain curves) are based on material properties.
- d. In case a numerical analysis (e.g. FE) is performed, all possible buckling mode shapes shall be considered in determining the minimum buckling load (As guide, API 579-1/ASME FFS-1, Annex B1, par. B1.4.2 can be used)

C2.5.6 Creep-fatigue assessment of dissimilar weld joints

Dissimilar welds are not present and hence not evaluated during this research. Also, the data available is not meant for nickel-based alloys.

C2.5.7 Microstructural approach

Microstructural approaches are generally used in conjunction with one another or to supplement other techniques, because these approaches are limited applicable and have a significant uncertainty. Therefore, decisions are commonly not made based on a single set of microstructural observations. Advantages of the microstructural approach are good availability, convenient in use and often not expensive. After being in service, critical areas should be sampled and a database should be constructed. Also, indications of time-dependent damage can be observed in changes over time in appearance, hardness, void size/population.

The remaining life of a component can be estimated on these changes:

- 1. Hardness and tensile strength are often correlating with remaining life. When the hardness is above a reference level, the material is considered suitable for continued service at nominal design stress levels and temperatures. The change in hardness gives an indication of damage accumulation over time.
- 2. The evaluation of creep cavities is in some alloys measure for crack initiation and failure. The time to the appearance of detectable creep cavities depends on factors like microstructure contents, grain size, stress, and heat treatment. The reported numbers, size and spacing of cavities depends on the microscopic techniques, sample preparation and experience of the observer. Several considerations for the creep cavity evaluation:
 - a. Indicative for creep damage larger than 50%: the presence of multiple cavities larger than 1.5µm on most grain boundaries normal to the principal stress or linked cavities appearing as micro-cracks or fissures.
 - b. Preparation of the samples must be adequate for cavity detection and sizing by optical and/or electron microscopic techniques (e.g. samples free of oxides)
 - c. Areas especially susceptible to cavitation damage:
 - i. Fine grain regions of the HAZ
 - ii. Regions of high stress concentrations
 - iii. Regions of coarse precipitate particles
 - d. The optical microscope may be used to obtain information about the creep resistance or damage state of the material. This information might be used to modify the strain rate and other coefficient equations (e.g. for the Omega method). Important to note is that an increase in grain size results commonly in a decrease of strain rate.

C2.6Remediation

In the situation that the component does not satisfy the creep damage criterion within the required service life or in case the sensitivity analysis indicates unacceptable results, then remedial action is required. The most likely action is a change in service parameters (e.g. load, temperature, service life) and then re-performing the assessment.

C2.7 In-service monitoring

Pressure and temperature are the most effective tools to monitor equipment subject to creep damage. The information obtained from the monitoring can be used to update the creep remaining life calculation to determine if continued operation is acceptable.

C2.8 Documentation

Documentation essential for the assessment has to be provided and stored with the equipment record files. At least included in this documentation is:

- a. Assessment level and any deviations to this level
- b. Loading conditions (e.g. load histogram) showing the assumed start-up, normal, upset, and shut-down conditions. Besides the primary loading conditions, additional loads should be addressed, like thermal gradients and residual stresses.
- c. Stress analysis results: the stress analysis method (i.e. FEM) and categorisation of the results.
- d. Material properties based on a material specification. Examples are yield strength, ultimate tensile strength, fracture toughness and creep properties. Also, registered should be the source of the data and if the data is obtained by direct testing or indirect. In addition, the process environment should be described including its effect on material properties. For cyclic loads, the interaction between creep and fatigue should be documented.
- e. Sensitivity analysis, which contains a listing of input parameters used to perform a sensitivity study (e.g. loads, service time, material properties). In case more sensitivity analyses are required, the results of each individual analysis shall be summarised.
- f. All assumptions used in the assessment should be documented and justified.
- g. All departures from the procedures described in the API 579-1/ASME FFS-1 should be reported and separately justified.
- h. All microstructural examination results shall be included in the documentation.

C3. Required tube thickness and maximum allowable working pressure

The API 579-1/ASME FFS-1 provided equations for minimum thickness and maximum allowable working pressure (MAWP) for straight sections of pipe and pipe bends subject to internal pressure.

The longitudinal stress (σ^L) is determined for $P \le 0.385 \cdot S \cdot E$ and $t_{min}^L \le 0.5R$, with P is internal design pressure, S is the allowable stress, E_q is quality factor from the code (=0.7 in case unknown), t_{min}^L is the minimum required thickness based on the longitudinal membrane stress for a tube, and R is the inside radius of the tube:

$t_{min}^L = \frac{P \cdot R}{2S \cdot E + 0.4P} + t_{sl}$,	(C5)	
$MAWP^{L} = \frac{2S \cdot E(t_c - t_{sl})}{R - 0.4(t_c - t_{sl})}$,	(C6)	and
$\sigma_m^L = \frac{P}{E_a} \left(\frac{R}{t_c - t_{sl}} - 0.4 \right)$		(C7)	

Here, MAWP^L is the maximum allowable working pressure based on longitudinal stress, and σ_m^L is the nominal longitudinal membrane stress for a tube. t_{sl} is the supplemental thickness for mechanical loads other than internal pressure such as weight of the tube itself or its contents, constraint of thermal expansion, temperature gradients. The critical thickness (t_c) is the result of:

$$t_c = t - \text{LOSS} - \text{FCA} \qquad , \tag{C8}$$

in which t is the nominal thickness of the tube, LOSS is the metal loss due to service conditions, and FCA is the specified future corrosion allowance. The FCA might be determined from corrosion rates determined in similar service conditions, from previous thickness measurements or from corrosion design curves.

Circumferential stress (σ^{C}) for thicknesses $\leq 0.5R$, together with the minimum required thickness (t_{min}^{C}) and the maximum allowable working pressure based on circumferential stress (MAWP^C) can be calculated using:

$$t_{min}^{C} = \frac{P \cdot D}{2S + P} + 0.005D + e_{t} , \qquad (C9)$$

$$MAWP^{C} = \frac{2S(t_{C} - 0.005D - t)}{D - (t_{C} - 0.005D - t)} , \qquad (C10) \text{ and}$$

$$\sigma^{C} = \frac{P[D - (t_{C} - 0.005D - e_{t})]}{2(t_{C} - 0.005D - e_{t})} . \qquad (C11)$$

Where D is the tube diameter and e_t is a parameter that is dependent on the presence of a weld and the geometry of that weld (e.g. strength of weld directly welded to the header or set into a tube seat).

C4. Stress analyses for protection against damage

The API 579-1/ASME FFS-1 provides procedures for performing stress analyses to determine protection against several damage mechanisms like plastic collapse, local failure, buckling, and cyclic loading. Using these procedures, necessary details are generated to obtain a consistent result in relation with the development of loading conditions, selection of material properties, comparison to acceptance criteria, and post-processing of results. To perform a proper stress analysis, data and material models are used (e.g. as described in API 579-1/ASME FFS-1, Annex F). Material data required is at least:

- 1. Physical properties Young's Modulus, Poisson's ratio, thermal expansion coefficient, thermal diffusivity, thermal conductivity, density.
- 2. Strength parameters Allowable stress, minimum yield strength, minimum tensile strength.
- 3. Monotonic stress-strain curve elastic perfectly plastic and elastic-plastic true stress-strain curve with strain hardening.
- 4. Cyclic stress-strain curve Stabilised true stress-strain amplitude curve.

C4.1 Histogram

In case (any of) the loads vary with time, a loading histogram has to be developed to show the time variation of each specific load. The loading histogram shall include all significant operating temperatures, pressures, supplemental loads, and exposure times. Also, the number of cycles (e.g. start-ups/shutdowns, normal operation, upset conditions), the anticipated sequence of operation, and the relation/interaction between the applied loadings must be considered. When an accurate histogram cannot be made, an approximate histogram should be generated. The information accompanying the approximate histogram include a description of all assumptions made and a discussion on the accuracy of the established points on the histogram. In conjunction, a sensitivity analysis must be performed to determine and evaluate the effects of the made assumptions.

C4.2 Protection against plastic collapse

Plastic collapse load is the load that causes overall structural instability. This is indicated by the inability to achieve an equilibrium of loads after a small load increase. For the evaluation of protection against plastic collapse, three analysis methods are given [C1]:

- 1. Elastic stress analysis method Allowable stress is that low to avoid plastic collapse.
- 2. Limit-load method A lower bound to the limit load is established by applying design factors to avoid the onset of gross plastic deformations (i.e. plastic collapse).
- 3. Elastic-plastic stress analysis method the collapse load is derived considering both applied loading and deformation characteristics. Design factors are applied to the plastic collapse load to establish the allowable load.

Due to the relative small wall thickness and the large temperature gradient around the circumferential of the boiler tube investigated, plastic deformation cannot be avoided and hence elastic stress analysis cannot be used. Some of the drawbacks of the limit-load method are the disregarding of temperature fields and reduction in resistance (i.e. weakening) due to deformation. This makes the limit-load method not suitable for thin-walled boiler tubes and its circumstances. Elastic-plastic (E-P) stress analysis method remains for the plastic collapse evaluation, especially since this method regards the stress redistribution and deformation characteristics occurring because of inelastic.

C4.2.1 Acceptance criteria

The protection against plastic collapse is achieved by satisfying two criteria [C1]:

- a. General criteria The plastic collapse load is determined in the E-P stress analysis in which the load combinations are evaluated by the concept of Load and Resistance Factor Design (LRFD). These factored loads include a design factor to account for uncertainty. The resistance of the component against these factored loads is then determined using the E-P stress analysis. These factors are for example given in API 579-1/ASME FFS-1, Table B1.4.
- b. Service criteria The design should incorporate the service criteria, which may limit the component's integrity. All locations in the component must withstand the design loads (e.g. see API 579-1/ASME FFS-1, Table B1.4). The effect of deformation on service performance has to be evaluated at the design load combinations. Although the plastic collapse criteria may be met, the deformation can be that excessive that design loads must be reduced to meet the deformation criteria. Some examples to be considered are piping connections and interference with other components.

C4.2.2 Assessment procedure

A component meets its protection against plastic collapse when using an E-P stress analysis, if the assessment procedure is followed:

- a. **STEP 1** FEM model must include all relevant geometries, loads and boundary conditions. In addition, model refinement is needed at and around areas of stress and strain concentrations.
- b. **STEP 2** All relevant loads and applicable load cases shall be defined. As minimum the API 579-1/ASME FFS-1, Tables B1.1, can be included.
- c. **STEP 3** Employing the E-P stress analysis: in case plasticity is expected, the Von Mises yield function and its associated flow rule must be used. A true stress- true strain curve model that includes temperature-dependent hardening should be provided.
- d. **STEP 4** Load case combination determination for the E-P stress analysis by using the information from STEP 2.
- e. **STEP 5** Performing E-P analysis for each of the load cases as defined in STEP 4. In case convergence is achieved, the component is stable under the applied loads. Otherwise, the applied loads have to be reduced or the wall thickness has to be increased, and the analysis has to be performed again.

C4.3 Stress analysis for protection against local failure

Adjacent to the protection against plastic collapse, local failure criteria must be satisfied. This is solely the case if the significance of a strain concentration cannot be established (i.e. the design code does not provide any guidance on this specific case). The E-P stress analysis is also the most accurate method for the estimation of the protection against local failure. The acceptance criteria are similar to those described for the protection against plastic collapse.

C4.3.1 Assessment procedure

The protection against local failure procedure can be constructed as shown underneath:

- a. STEP 1 E-P stress analysis performance based on load case combinations for the local criteria (like those given in API 579-1/ASME FFS-1, Table B1.4). Non-linear geometry effects should be considered in the analysis.
- b. **STEP 2** For the location, the principal stresses (i.e. σ_1 , σ_2 , and σ_3) have to be determined, together with the equivalent stress (σ_e) and the total equivalent plastic strain (ε_{peq}). The equivalent stress (σ_e) is similar to the Von Mises equivalent stress:

$$S = \sigma_e = \frac{1}{\sqrt{2}} [(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2] \quad . \tag{C12}$$

c. **STEP 3** – Determination of the triaxial strain limit (ε_L):

$$\varepsilon_L = \varepsilon_{Lu} \cdot \exp\left[-\left(\frac{\alpha_{sl}}{1+m_2}\right)\left(\left\{\frac{(\sigma_1+\sigma_2+\sigma_3)}{3\sigma_e}\right\} - \frac{1}{3}\right)\right] \quad , \tag{C13}$$

In which ε_{Lu} (uniaxial strain limit), m_2 (i.e. strain-hardening exponent), and α_{sl} (material factor for the multi-axial strain limit), are determined from API 579-1/ASME FFS-1, Table B1.6. For Alloy 617, a nickel-based Super Alloy, the following equations are valid:

$$\begin{split} m_2 &= 1.90(0.93 - R_{\sigma}) , \qquad (C14) \\ \varepsilon_{spec} &= \ln \left[1 + \frac{\varepsilon}{100} \right] , \qquad (C15) \\ RA_{spec} &= \ln \left[\frac{100}{100 - RA} \right] , \qquad (C16) \text{ and} \\ \alpha_{sl} &= 2.2 . \qquad (C17) \end{split}$$

Here R_{σ} is the ratio of the minimum yield strength to the minimum ultimate tensile strength, ε_{spec} is the specified elongation, ε is the elongation (in %), RA_{spec} is the specified reduction in area and RA is the reduction in area (in %). In case the elongation and the reduction in area are not specified, then $\varepsilon_{Lu} = m_2$. When the elongation and reduction in area are specified:

$$\varepsilon_{Lu} = \max[m_2, \varepsilon_{spec}, RA_{spec}] \qquad . \tag{C18}$$

- d. **STEP 4** Determination of the forming strain (ε_{cf}) based on the material and fabrication method in according with the applicable design code. If heat treatment is performed in according with that design code, the forming strain may be assumed to be zero.
- e. **STEP 5** Verify if the strain limit is satisfied:

$$\varepsilon_{peq} + \varepsilon_{cf} \le \varepsilon_L \tag{C19}$$

In lieu of the assessment procedure, the evaluation of a specific loading sequence requires a strain limit damage calculation procedure. For that, the loading path has to be divided into load increments and each load increment has to be evaluated separately using the equations as described in the assessment procedure. The strain limit damage (D_{ε}) is then calculated per load increment.

C4.4 Stress analysis for protection against collapse from buckling

Two options exist to account for protection against collapse from buckling [C1]:

- a. In the situation that a E-P stress analysis is performed to determine the protection against plastic collapse, the buckling factor is accounted for when using the load combinations in API 579-1/ASME FFS-1, Table B1.4.
- b. A bifurcation buckling analysis can be performed by E-P analysis. This is done with the effects of non-linear geometry in the solution to determine the pre-stress in the component. In this case, a minimum buckling factor of $\Phi_{\rm b} = \frac{1.667}{\beta_{cr}}$ has to be used (with $\Phi_{\rm b}$ is the buckling factor and β_{cr} is the buckling reduction factor). The pre-stress can be determined based on the loading combinations in API 579-1/ASME FFS-1, Table B1.2. The buckling reduction factor (β_{cr}) may be used to account for imperfections in the component, but other factors can be used when developed from published data. The β_{cr} for a cylinder (i.e. tube) under axial compression with an outer diameter (D_o) to wall thickness (t) ratio of approximately 20:

$$\beta_{cr} = \frac{338}{389 + \frac{D_0}{t}} = \frac{338}{409} = 0.83 \tag{C20}$$

C4.5 Stress analysis for protection against failure from cyclic loading (fatigue)

Thermal cycling of the boiler tubes causes fatigue and a fatigue evaluation must be performed to determine the protection against failure from cyclic loading. First, screening criteria have to be investigated to determine if a fatigue analysis is required. In case a fatigue evaluation is required, fatigue curves may be used (if applicable) and ratchetting has to be taken into account.

C4.5.1 Screening criteria for fatigue

Fatigue analysis is not required when the provisions of one of three options are met. The first option is based on experience with comparable equipment. This experience is not (sufficient) available and this option is for that reason disregarded. Method A is the second option and can solely be used for materials with a specified minimum ultimate tensile strength \leq 552 MPa. Nickel-based alloy 617 has a specified minimum ultimate tensile strength of 665 MPa, and therefore method A is not valid. Method B remains as option and can be used for all materials [C1]:

- a. STEP 1 Load history determination including all significant cyclic operation loads and events (see paragraphs 2.6.1 and 2.6.2).
- b. **STEP 2** Determine the fatigue screening criteria factors C_1 and C_2 based on the type of construction (e.g. in accordance with API 579-1/ASME FFS-1, Table B1.9, and ASME VIII, div. 2, Table 5.10). The boiler tube is considered being an "integral structure", since it absorbs all or most of the stresses applied by itself (i.e. the tube is not able to transfer most/all of the stresses to the structural steel). The tube is also considered a "flawless other component". The adherent values are $C_1 = 3$ and $C_2 = 2$.
- c. **STEP 3** The design number of full-range pressure cycles $(N_{\Delta FP})$ can be calculated based on the load histogram (STEP 1). In case the following equation is valid, continue to STEP 4:

$$N_{\Delta FP} \le N(C_1 S_m) \tag{C21}$$

in which S_m is the allowable stress based on the material and design temperature. $N(C_1S_m)$ is the number of cycles until failure for C_1S_m (i.e. $3S_m$).

d. **STEP 4** – Determination of the maximum range of pressure fluctuation during normal operation (ΔP_N) and the adjacent number of significant cycles ($N_{\Delta P}$) based on the load histogram of STEP 1. In which S_{as} is the stress amplitude taken from the fatigue curve for a certain temperature evaluated at 10⁶ cycles. Start-ups and shut-downs are not considered here. In case the following equation is satisfied, proceed to STEP 5.

$$\Delta P_N \le \frac{P}{C_1} \left(\frac{S_a(N_{\Delta P})}{S_m} \right) \tag{C22}$$

In which P is the design pressure of the system, and S_a is alternating stress obtained after a specified number of operating cycles.

e. **STEP 5** – Determination of the maximum temperature difference between any two adjacent points during normal operation (ΔT_N) and the corresponding number of cycles ($N_{\Delta TN}$) based on the load histogram from STEP 1. In case the following equation is satisfied, go to STEP 6.

$$\Delta T_N \le \left(\frac{S_a(N_{\Delta TN})}{C_2 E_{ym} \alpha_m}\right) \tag{C23}$$

in which E_{ym} is Young's modulus at the mean temperature of the cycle, and α_m is the coefficient of thermal expansion at the mean temperature.

f. **STEP 6** – Determination of the maximum range of temperature difference fluctuation (ΔT_R) during normal operation excluding start-ups and shutdowns and the corresponding number of significant cycles ($N_{\Delta TR}$) based on the load histogram from STEP 1. In case the following equation is satisfied, proceed to STEP 7.

$$\Delta T_R \le \left(\frac{S_a(N_{\Delta TR})}{C_2 E_{ym} \alpha_m}\right) \tag{C24}$$

- g. STEP 7 Determination of the maximum range of temperature difference fluctuation during normal operation for two adjacent components manufactured from different materials. Only Alloy 617 is considered in this research and for that reason this step is not applicable.
- h. **STEP 8** Determination of the equivalent stress range derived from the specified full range of mechanical loads (ΔS_{ML}) including piping reactions, but excluding pressure and corresponding number of significant cycles ($N_{\Delta S}$) based on the load histogram from STEP 1. A fatigue analysis is not required in case:

$$\Delta S_{ML} \le S_a(N_{\Delta S}) \tag{C25}$$

C4.5.2 Assessment procedure

A fatigue analysis is performed using an elastic-plastic (E-P) method. The effective strain range is calculated for each cycle in the load histogram using either cycle-by-cycle analysis or the Twice Yield method. The calculated results are used to evaluate the fatigue damage. For the cycle-by-cycle analysis, a cyclic plasticity algorithm with kinematic hardening shall be used. The Twice Yield Method is based on a specified stabilised cyclic stress range-strain range and the strain range. Advantage of this method is that it is performed in the same manner as a monotonic analysis and does not require unloading and reloading as with the cycle-by-cycle analysis. The Twice Yield Method may be used with an analysis program without cyclic plasticity capability. The stabilised cyclic stress-strain curve and other material properties are based on the data at average temperature of the cycle. This data can be obtained by testing of by using the data from other sources.

- a. STEP 1 Load histogram determination (for example with the methods in API 579-1/ASME FFS-1, Annex B2).
- b. STEP 2 Determination of the individual stress-strain cycles for a location subject to fatigue (by using the cyclic counting methods in API 579-1/ASME FFS-1, Annex B2). The total number of cyclic stress ranges in the histogram is defined as *M*.
- c. STEP 3 Determination of the loading at the start and end point of the kth cycle counted in STEP 2. The loading ranges (i.e. differences between the loadings at the start and end point of the cycle) have to be determined based on this data.
- d. STEP 4 E-P stress analysis shall be performed for the kth cycle. For cycle-to-cycle analysis, constant amplitude loading is cycled using cyclic stress amplitude-strain amplitude curve. For the Twice Yield Method, the loading at the start point of the cycle is zero and the loading at the end point is the loading range determined in STEP 3. For the analysis, the cyclic stress range-strain range curve or hysteresis loop stress-strain curve can be used. For thermal loading, the loading range in the Twice Yield Method may be applied by specifying the temperature field at the start point of the cycle as an initial condition and applying the temperature field at the end point of the cycle in a single loading step.
- e. **STEP 5** Calculation of the effective strain range ($\Delta \varepsilon_{eff,k}$) for the kth cycle:

$$\Delta \varepsilon_{eff,k} = \frac{\Delta S_{P,k}}{E_{ya,k}} + \Delta \varepsilon_{peq,k} \qquad , \tag{C26}$$

where $\Delta S_{P,k}$ is the range of primary plus secondary plus peak equivalent stress for the kth cycle (i.e. the Von Mises equivalent stress range), $E_{ya,k}$ is the value of elasticity modules of the material at the point under consideration, evaluated at the mean temperature of the kth cycle, and $\Delta \varepsilon_{peq,k}$ is the equivalent plastic strain range for the kth loading cycle. The $\Delta S_{P,k}$ and the $\Delta \varepsilon_{peq,k}$ are calculated:

$$\Delta S_{P,k} = \frac{1}{\sqrt{2}} \sqrt{\frac{\left(\Delta \sigma_{11,k} - \Delta \sigma_{22,k}\right)^2 + \left(\Delta \sigma_{11,k} - \Delta \sigma_{33,k}\right)^2 + \left(\Delta \sigma_{22,k} - \Delta \sigma_{33,k}\right)^2 + 6\left(\Delta \sigma_{12,k}^2 + \Delta \sigma_{13,k}^2 + \Delta \sigma_{23,k}^2\right)}{6\left(\Delta \sigma_{12,k}^2 + \Delta \sigma_{13,k}^2 + \Delta \sigma_{23,k}^2\right)}, \quad (C27) \quad \text{and}$$

$$\Delta \varepsilon_{peq,k} = \frac{\sqrt{2}}{3} \sqrt{\frac{\left(\Delta p_{11,k} - \Delta p_{22,k}\right)^2 + \left(\Delta p_{22,k} - \Delta p_{33,k}\right)^2 + \left(\Delta p_{33,k} - \Delta p_{11,k}\right)^2 + 1.5\left(\Delta p_{12,k}^2 + \Delta p_{23,k}^2 + \Delta p_{31,k}^2\right)}, \tag{C28}$$

in which $\Delta \sigma_{ij}$ is the stress component, Δp_{ij} is the plastic strain range for the kth cycle, and 'i' and 'j' are the stress directions. Often the $\Delta S_{P,k}$ and the $\Delta \varepsilon_{peq,k}$ are output variables obtained directly from the stress analysis.

f. **STEP 6** – Determination of the effective alternating equivalent stress ($S_{alt,k}$) for the kth cycle:

$$S_{alt,k} = \frac{1}{2} E_{yf} \cdot \Delta \varepsilon_{eff,k} \qquad , \qquad (C29)$$

in which E_{yf} is the value of the elasticity modulus of the fatigue curve. For this value the Young's modulus at the assessment temperature or at the mean temperature can be taken.

g. **STEP 7** – Determination of the permissible number of cycles (N_k) for the alternating equivalent stress as calculated at STEP 6. Fatigue curves can be created out of testing or from the fatigue curves as given in other sources.

h. **STEP 8** – Compute the fatigue damage for the kth cycle $(D_{f,k})$ with the actual number of repetitions of the kth cycle (n_k) .

$$D_{f,k} = \frac{n_k}{N_k} \tag{C30}$$

- i. **STEP 9** Repeating STEPs 3 to 8 for all stress ranges (*M*) identified in the cycle counting process of STEP 2.
- j. **STEP 10** Calculation of the accumulated fatigue damage by summation of the value computed at STEP 9. The location in the component is acceptable in the protection against failure from cyclic loading in case:

$$\sum_{k=1}^{M} D_{f,k} \le 1.0$$
 (C31)

k. **STEP 11 –** Repeat STEPs 2 to 10 for each location in the component subject to fatigue.

The fatigue assessment of welds is also handled, though at the moment welds are not considered in the research.

C4.5.3 Ratchetting Assessment

The elastic-plastic (E-P) stress analysis can be used to evaluate the protection against ratchetting. This assessment is performed in a similar manner as that for fatigue, though is limited to thermal load cycling. In case protection against ratchetting is satisfied, it is assumed that progression of the stress-strain hysteresis loop along the strain axis cannot be sustained with cycles and that the hysteresis loop will stabilise. As result a separate check for plastic shakedown to alternating plasticity is not required.

- a. **STEP 1** Numerical model set-up including all relevant geometry characteristics and boundary conditions.
- b. STEP 2 Definition of all relevant loads including internal maximum allowable working pressure, weight of the remainder material (dead weight) underneath the location evaluated, effects of steady-state and transient fluid momentum, and the self-restraining load for thermal loads or applied displacements.
- c. STEP 3 Utilisation of the Von Mises yield function and associated flow rule. The plastic limit is defined by the minimum yield strength at temperature. In the stress analysis, non-linear geometry effects must be considered.
- d. **STEP 4** For the (combination of) load(s) that causes the highest likelihood of ratchetting, an E-P stress analysis shall be performed for a number of cycling repetitions.
- e. STEP 5 After application of at least three complete cycling repetitions, the ratchetting criteria are evaluated. If any of the following conditions is achieved, the ratchetting criteria are satisfied. In case these conditions are not met, the wall thickness of the boiler tube has to be increased or the applied loads have to be reduced. Additional cycles may be needed in the E-P stress analysis to ensure convergence of the results (i.e. stabilisation of the stress-strain hysteresis loop). The criteria/conditions:
 - 1. No plastic strains in the component
 - 2. An elastic core in the primary-load-bearing boundary exists
 - 3. The overall dimensions of the component do not change permanently. This can be shown by plotting the relevant dimensions versus time between the last and the before-last cycles.

C4.5.4 Fatigue analysis: cycle counting

The API 579-1/ASME FFS-1 (edition 2007), Annex B3, provides two cycle counting methods, though an alternative cycle counting method may be used.

The rainflow method

The rainflow cycle counting method according to ASTM E1049 is recommended to determine the time points representing individual cycles in case the stress or strain variations can be expressed in a single parameter. This method is not suitable for non-proportional loading. Due to the thermal variations (i.e. fluctuating thermal gradient) around the circumferential of the boiler tube during a thermal cycle, the load orientations cannot be considered fixed. This causes a non-proportional loading, and hence the rainflow method is not suitable.

Max-Min cycle counting method

In case of non-proportional loading, the max-min cycle counting method is recommended to determine the time points representing individual cycles. This is the method used in the project and already described in paragraph 2.5.4 of the main report.

C5. Material properties for a FFS assessment

This section provides a guidance on the materials information required for the Fitness-For-Service (FFS) assessment. When incorporating the material properties, the assessment should reflect any change or degradation (e.g. due to aging).

C5.1 Yield strength and ultimate tensile strength

The most proper manner to obtain yield strength (YS) and ultimate tensile strength (UTS) is by tensile testing the material in accordance with a standard, like ASTM A370 or ASTM E8. In case the temperature used for the FFS assessment is significant different than the temperature at which the yield strength and ultimate tensile strength are determined, a temperature correction factor has to be applied. For nickel-based alloys, like Alloy 617, the yield strength and ultimate tensile strength cannot be converted from the hardness. In the absence of heat specific data, mean values for the yield strength and the ultimate tensile strength can be estimated by:

$\sigma_{uts}^{mean} = \sigma_{uts}^{min} + 69 MPa$,	(C31)	and
$\sigma_{vs}^{mean} = \sigma_{vs}^{min} + 69 MPa$		(C32)	

For Alloy 617 the yield strength and ultimate tensile strength at several temperatures are provided in a number of material data sources.

C5.2 Flow stress

The flow stress (σ_f) may be referred to as the effective yield strength of a work hardened material. The concept of the flow stress allows a material to be treated like an elastic-plastic material that can be characterised by one strength parameter. For example, the flow stress can be used as the stress level in the material that controls the resistance against plastic collapse failure. The flow stress can be estimated by one of these equations:

$$\sigma_f = \frac{(\sigma_{ys} + \sigma_{uts})}{2} \qquad , \tag{C33}$$
$$\sigma_f = \sigma_{ys} + 69 MPa \qquad , \qquad (C34) \quad \text{or}$$

$$\sigma_f = \frac{\sigma_{ys}}{2} \left[1 + \frac{\left(\frac{n_{RO}}{0.002}\right)^{n_{RO}}}{\exp(n_{RO})} \right] \qquad . \qquad (C35)$$

C5.3 Material Properties Council model

The model predictive control (MPC) model can be used in FFS calculations for constructing a stressstrain curve in case strain-hardening characteristics must be considered. This model is described in paragraph 4.3 of the report.

C5.3.1 Tangent modulus in the MPC model

The slope of the stress-strain curve at any specified stress or strain is described by the tangent modulus (E_t). Before the proportional limit, the tangent modulus is equal to the Young's modulus. Beyond the proportional limit, the tangent modulus varies with the strain. In the MPC model, the tangent modulus beyond the proportional limit is equated:

$$E_t = \frac{\partial \sigma}{\partial \varepsilon_t} = \left(\frac{\partial \varepsilon_t}{\partial \sigma}\right)^{-1} = \left(\frac{1}{E_y} + D_1 + D_2 + D_3 + D_4\right)^{-1}$$
(C36)

At which D_1 to D_4 are coefficients for the calculation of the tangent modulus. These are calculated:

$$D_{1} = \frac{\sigma^{\left(\frac{1}{m_{1}}-1\right)}}{2m_{1}A_{1}^{\left(\frac{1}{m_{1}}\right)}}, \qquad (C37)$$

$$D_{2} = -\frac{1}{2} \left(\frac{1}{A_{1}^{\left(\frac{1}{m_{1}}\right)}}\right) \cdot \left(\sigma^{\left(\frac{1}{m_{1}}\right)} \left\{\frac{2}{\kappa(\sigma_{uts} - \sigma_{ys})}\right\} \{1 - \tanh^{2}[H]\} + \frac{1}{m_{1}} \sigma^{\left(\frac{1}{m_{1}}-1\right)} \tanh[H]\right), \qquad (C38)$$

$$D_3 = \frac{\sigma^{\left(\frac{1}{m_2} - 1\right)}}{2m_2 A_2^{\left(\frac{1}{m_2}\right)}} \hspace{1.5cm} \text{,} \hspace{1.5cm} \text{(C39) and}$$

$$D_{4} = \frac{1}{2} \left(\frac{1}{A_{2}^{\left(\frac{1}{m_{2}}\right)}} \right) \cdot \left(\sigma^{\left(\frac{1}{m_{2}}\right)} \left\{ \frac{2}{K(\sigma_{uts} - \sigma_{ys})} \right\} \{1 - \tanh^{2}[H]\} + \frac{1}{m_{2}} \sigma^{\left(\frac{1}{m_{2}} - 1\right)} \tanh[H] \right)$$
(C40)

The other symbols are the same as for the MPC model as described in paragraph 4.3 of the report.

C5.4 Ramberg-Osgood Model

Another model to construct a stress-strain curve for FFS assessment calculations, is the Ramberg-Ostgood (RO) model. In case a J-integral calculation is required, the exponent (n_{RO}) can be used. The true strain (ε_{ts}) is computed by:

$$\varepsilon_{ts} = \frac{\sigma_t}{E_y} + \left(\frac{\sigma_t}{H_{RO}}\right)^{\frac{1}{n_{RO}}} = \ln(1 + \varepsilon_{es}) , \qquad (C41) \text{ and}$$
$$\sigma_t = (1 + \varepsilon_{es})\sigma_{es} , \qquad (C42)$$

in which ε_{es} and σ_{es} are the engineering strain and engineering stress respectively, n_{RO} is an material exponent and H_{RO} is a constant.

In case multiple data points are available for stress-strain curve construction, the data fitting exponent n_{RO} and constant H_{RO} can be determined using regression techniques. When solely the yield and ultimate tensile strength are known, n_{RO} and H_{RO} can be calculated:

n _{RO}	$- \frac{1+1.3495(R)-5.3117(R)^2+2.9643(R)^3}{2}$				(CA2)	with
	1.12 .0097(.	$(R) = 11.7464(R)^2$,	(C43)	WILII
<i>R</i> =	$\frac{\sigma_{ys}}{\sigma_{uts}}$,		(C44)	and
H _{RO}	$=\frac{\sigma_{uts}\exp[n_{RO}]}{n_{RO}^{n_{RO}}}$				(C45)	

C5.5 Cyclic stress strain curve

The cyclic stress-strain curve (i.e. strain amplitude versus stress amplitude) can be calculated by equation (4.5A) and must always be part of the fit-for-service assessment.

C5.6 Physical Properties

The design codes WRC 503 or the ASME B&PV Code, Section II, Part D, and other data sources are providing physical property values for the full temperature range of:

- Young's modulus (E_y)
- Poisson's ratio (v)
- Coefficient of thermal expansion (α)
- Thermal conductivity (for thermal stress calculation)
- Thermal diffusivity (for transient thermal stress calculation)
- Density

C5.7 Fracture toughness

The fracture toughness is the ability of the material to withstand crack initiation and propagation. The FFS assessment for the fracture toughness can be based upon the critical stress intensity factor (K_{IC}), the critical value of the J integral (J_{crit}), and/or the critical crack tip opening displacement (CTOD or δ_{crit}). In case this data is available, it is usually based upon components containing crack-like flaws. Although solely components without any flaws are acceptable for this research (since the crack-like flaws are not acceptable due to the relatively small tube wall thickness in combination with a cyclic loading), it is valuable to determine the resistance of the component versus irregularities at the surface (e.g. scratches) that may cause stress concentrations.

C5.7.1 Assessment of fracture toughness from Charpy V-Notch data

The fracture toughness (in K_{IC}) can be obtained directly from the Charpy V-Notch (CVN) impact energy using correlations. Therefore, the impact energy should be determined at the lowest temperature expected for the component (i.e. most brittle circumstances). For the research that would be CVN testing at room temperature (i.e. 20 °C). Commonly an impact energy value of 27 Joule is an acceptable value for fracture toughness/impact strength when tested using a standard sized CVN sample. Effects of a reduced cross sectional area of a sub-sized CVN sample (due to the thinner wall thickness of the boiler tube) on the absorbed energy and the transition temperature must be considered by using correlations like provided in the ASTM A370. Assessing fracture toughness (K_{IC}) from CVN results for the boiler tubes, upper shelf correlations can be used, since the high service temperatures and FCC microstructure of Alloy 617 make it likely that brittle fracture is avoided. The first correlation is from Rollfe-Novak-Barsom:

$$K_{IC} = 0.8\sigma_y \sqrt{\left(\frac{CVN}{\sigma_y} - 0.01\right)}$$
(C46)

The second correlation is based on the lower bound estimate recommended in the WRC 265 standard:

$$K_{IC} = \sigma_{ys} \sqrt{0.52 \left(\frac{CVN}{\sigma_{ys}} - 0.02\right)}$$
(C47)

Two cautions must be regarded for establishing the fracture toughness based on CVN data [C1]:

- 1. The CVN data should be representative for the component being evaluated (i.e. should be heat and heat treatment specific and should reflect the microstructure).
- 2. An appropriate temperature must be chosen to perform the assessment. The worst-case loading may not be at the lower temperature, but at a higher temperature.

C5.8 Material data for creep analysis

C5.8.1 Time to rupture

An estimation of the time to rupture can be obtained by the method as described in paragraph 2.4 of the report or by using either one of two methods: Material Properties Council (MPC) Project Omega program or the API STD 530 data, which uses the Larson-Miller Parameter (LMP).

MPC Project Omega Data

In the MPC method, the rate of strain accumulation is predicted by using a strain-rate parameter and a multi-axial damage parameter.

$$L = \frac{1}{\dot{\varepsilon}_{co}\Omega_m}$$
 (C48)

in which L is the remaining life, $\dot{\varepsilon}_{co}$ is the initial creep strain rate at start of the time period based on stress state and temperature and parameter Ω_m is the Omega multi-axial damage parameter. The $\dot{\varepsilon}_{co}$ can be equated:

$${}^{10}\log\dot{\varepsilon}_{co} = -\left\{ (A_0 + \Delta_{\Omega}^{sr}) + \left(\frac{1}{492 + 1.8T}\right) \left(A_1 + A_2S_l + A_3S_l^2 + A_4S_l^3\right) \right\}$$
(C49)

in which A_0 to A_4 are curve-fit coefficients for the MPC strain rate parameter (to be determined for Alloy 617), Δ_{Ω}^{sr} is an adjustment parameter for the creep strain rate which is -0.5 for the bottom of the scatter band and +0.5 for the top of the scatter band, T is the temperature in degrees Celsius and S_l is the logarithm for the effective stress.

$$S_l = {}^{10}\log(\sigma_e)$$
 , (C50) with

$$\sigma_e = \frac{1}{\sqrt{2}}\sqrt{(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2} , \qquad (C51)$$

in which σ_1 , σ_2 and σ_3 are principal stresses, and σ_e is the effective stress.

The Omega multi-axial damage parameter (Ω_m) can be equated:

$$\Omega_m = \Omega_n^{\delta_\Omega + 1} + \alpha_\Omega n_{BN} \qquad , \tag{C52}$$

in which Ω_n is the omega damage parameter based on uniaxial loading, δ_Ω is the damage parameter exponent, α_Ω is a parameter based on the state-of-stress (i.e. 3.0 for pressurised spheres or formed heads, 2.0 for pressurised cylinders of cones, and 1.0 for all other components and stress states), and n_{BN} is the Bailey Norton coefficient evaluated at the reference stress in a load increment.

$$\Omega_n = \max[(\Omega - n_{NB}), 3.0] \qquad , \qquad (C53) \quad \text{and} \quad$$

$${}^{10}\log\Omega = \left\{ \left(B_0 + \Delta_{\Omega}^{cd} \right) + \left(\frac{1}{492 + 1.8T} \right) \left(B_1 + B_2 S_l + B_3 S_l^2 + B_4 S_l^3 \right) \right\}$$
(C54)

Here, Ω is the Omega uniaxial damage parameter, and Δ_{Ω}^{cd} is an adjustment factor for creep ductility for a range of +0.3 for brittle behaviour and -0.3 for ductile behaviour. B_0 to B_4 are coefficients for the Omega parameter which still have to be determined for Alloy 617.

$$\delta_{\Omega} = \beta_{\Omega} \left(\frac{\sigma_1 + \sigma_2 + \sigma_3}{\sigma_e} - 1.0 \right) \tag{C55}$$

in which β_{Ω} is the Prager factor of 0.33. The Bailey Norton coefficient (n_{NB}) then becomes:

$$n_{NB} = -\left\{ \left(\frac{1}{492+1.8T} \right) \left(A_2 + 2A_3S_l + 3A_4S_l^2 \right) \right\}$$
[°C]
. (C56)

Drawback of the MPC method is that primary creep effects are not considered. The effects of primary creep may be neglected when the stress from the applied load is \leq 50% of the minimum yield strength at the assessment temperature.

API STD 530 Data

Different method to calculate the Larson-Miller parameter (LMP) ed:

 $LMP(\sigma) = (1.8T + 492) (C_{LMP} + {}^{10}log[L]) 10^{-3}$. [°C] (C57)

C5.8.2 Creep strain rate data

The creep strain rate ($\dot{\varepsilon}_c$) can be calculated by the method as described in paragraph 2.3.

C6. Literature

[C1] API 579-1/ASME FFS-1, *"Fitness-For-Service"*, American Society of Mechanical Engineers and American Petroleum Institute, June, 2007.

Appendix D – Mathematics and explanations

In this appendix, the mathematics and more detailed explanations are given for the theory in the report.

D1. Example calculation: Creep strain rate determination at 750 °C

As an example, the values of n and B_c for the creep rate at 750 °C have been determined. Equation (2.3D) can be rewritten as:

 $\log \dot{\varepsilon}_c = \log B_c + n \log \sigma - n \log \mu + \log \left(\exp \left(\frac{-Q_c}{RT} \right) \right)$, (D1A)

with:

and:

$$\log\left(\exp\left(\frac{-Q_c}{RT}\right)\right) = \frac{\ln\left(\exp\left(\frac{-Q_c}{RT}\right)\right)}{2.303} = \frac{-Q_c}{2.303RT} \quad , \tag{D1B}$$

In equation (D1A), B_c and n are unknown, while $\dot{\varepsilon}_c$ and σ can be determined from the graph shown in figure 2.12 of the report. Since most parameters are independent on the stress and remain unchanged for a certain temperature, the following deduction can be made:

$$\log \dot{\varepsilon}_{c,1} = \log B_c + n \log \sigma_1 - n \log \mu - \frac{Q_c}{2.303RT}$$
(D1C1)
$$\log \dot{\varepsilon}_{c,1} = \log B_c + n \log \sigma_c - n \log \mu - \frac{Q_c}{2.303RT}$$
(D1C2)

$$\frac{\log \varepsilon_{c,2} = \log B_c + n \log \sigma_2 - n \log \mu - \frac{1}{2.303RT}}{\log \dot{\varepsilon}_{c,1} - \log \dot{\varepsilon}_{c,2} = n \log \sigma_1 - n \log \sigma_2}$$
(D1C2)

which can be rewritten as:

$$\log \frac{\dot{\varepsilon}_{c,1}}{\dot{\varepsilon}_{c,2}} = n \log \frac{\sigma_1}{\sigma_2} \qquad , \tag{D1E}$$

Two points have been selected along the 750 °C trendline at:

σ_1	= 90 MPa	$\dot{\varepsilon}_{c,1}$	= 10 ⁻¹⁰ s ⁻¹
σ_2	= 480 MPa	$\dot{\varepsilon}_{c,2}$	= 10 ⁻⁴ s ⁻¹

Filling these points in equation (D1D), results in a value for the stress exponent of n = 8.3 at 750 °C. Now, the value of B_c can be determined by rewriting equation (2.3D) from the report:

$$B_c = \dot{\varepsilon}_c \left(\frac{\sigma}{\mu}\right)^{-n} \exp\left(\frac{Q_c}{RT}\right) \tag{D1F}$$

Using the parameters known, both calculated and determined from the graph, a value of B_c is obtained of 2.64 $\cdot 10^{33}$ s⁻¹. With the B_c obtained, the creep rate can be determined for each stress applied at 750 °C.

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The initial applied strain (ϵ_0) is converted to creep strain ($\epsilon_{c,s}(t)$) during stress relaxation and the resulting stress at time t is $\sigma(t)$:

$$\sigma(t) = E \cdot \left(\epsilon_0 - \epsilon_{c,s}(t)\right) \qquad , \tag{D2E}$$

with which equation (D2D) can be transferred to a function of strain over time ($\epsilon(t)$):

$$f(\epsilon(t)) = c_4 \cdot E^{c_2} \cdot \left(\epsilon_0 - \epsilon_{c,s}(t)\right)^{c_2} \qquad . \tag{D2F}$$

Since *E* is temperature-dependent but time independent, c_4 and E^{c_2} can be combined into c_5 and equation (D2A) becomes:

$$\dot{\epsilon} = c_5 \cdot \left(\epsilon_0 - \epsilon_{c,s}(t)\right)^{c_2} \tag{D2G}$$

The strain rate is the creep strain rate, so $\dot{\epsilon} = \dot{\epsilon}_{c,s}(t)$ and equation (D2G) is separable:

$$\dot{\epsilon}_{c,s}(t) \cdot \left(\epsilon_0 - \epsilon_{c,s}(t)\right)^{-c_2} = c_5 \qquad . \tag{D2H}$$

Integration of equation (D2H) over time results into:

$$\int \left(\epsilon_0 - \epsilon_{c,s}(t)\right)^{-c_2} \cdot d(\epsilon_{c,s}(t)) = -\left[\frac{\left(\epsilon_0 - \epsilon_{c,s}(t)\right)^{1-c_2}}{1-c_2}\right] = c_5 \cdot t + c_6', \tag{D2I}$$

in which c'_6 is the constant evolving from the integration. Equation (D2I) has to be written to the form where $\epsilon_{c,s}(t)$ is expressed in terms of the other parameters. First sub-step:

$$(\epsilon_0 - \epsilon_{c,s}(t))^{1-c_2} = (1-c_2)(c_6 - c_5 \cdot t)$$
 , (D2J)

with c_6 being the final constant evolving from the integration (negative of c'_6). This leads to:

$$(\epsilon_0 - \epsilon_{c,s}(t)) = ((1 - c_2)(c_6 - c_5 \cdot t))^{\frac{1}{1 - c_2}}$$
, (D2K)

D2. Stress relaxation: the analytic method.

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The development of creep strain in combination with relaxation stress may be described using an analytical method. First, the creep strain rate is expressed as a function of the stress, which in turn is dependent on the strain.

$$\dot{\epsilon} = f(\sigma(\epsilon))$$
 , (D2A)

Using the simplified Norton's creep equation as is used in ANSYS as well:

with,

 $\sigma(\epsilon) = \sigma(\epsilon_0 - \epsilon)$

 $f(\sigma) = c_4 \cdot \sigma^{c_2}$

 $f(\sigma) = c_1 \cdot \sigma^{c_2} \cdot e^{-c_3 T_K^{-1}}$

in which c_1 , c_2 and c_3 are temperature-dependent but time-independent parameters. The first and the last terms in equation (D2C) are time-independent and can be combined into parameter c_4 :

(D2B)

(D2C)

(D2D)

from which the $\epsilon_{c,s}(t)$ is singled out:

$$\epsilon_{c,s}(t) = \epsilon_0 - \left((1 - c_2)(c_6 - c_5 \cdot t) \right)^{\frac{1}{1 - c_2}}$$
 (D2L)

The value of c_6 is determined by using the boundary condition that for t = 0 no creep strain has developed yet (i.e. $\epsilon_{c,s}(t) = 0$):

$$0 = \epsilon_0 - \left((1 - c_2)c_6 \right)^{\frac{1}{1 - c_2}} \qquad \text{for } t = 0 \; . \tag{D2M}$$

The value of c_6 is then calculated as:

$$c_6 = \frac{\epsilon_0^{(1-c_2)}}{1-c_2}$$
 . (D2N)

Combining equations (D2L) and (D2N) gives the final equation for the analytic equation for the creep strain development over time:

$$\epsilon_{c,s}(t) = \epsilon_0 - \left((1 - c_2) \left(\frac{\epsilon_0^{(1 - c_2)}}{1 - c_2} - c_5 \cdot t \right) \right)^{\frac{1}{1 - c_2}} .$$
 (D20)

Knowing that $c_2 = n$, and:

$$c_5 = c_1 \cdot e^{-c_3 T_K^{-1}} \cdot E^{c_2}$$
 , (D2P)

results in:

$$\epsilon_{c,s}(t) = \epsilon_0 - \left((1-n) \left(\frac{\epsilon_0^{(1-n)}}{1-n} - c_1 \cdot e^{-c_3 T^{-1}} \cdot E^n \cdot t \right) \right)^{\frac{1}{1-n}} , \qquad (D2Q)$$

in which c_1 is:

$$c_1 = \frac{B_c}{\mu^n} \qquad , \qquad (D2R)$$

with B_c is a temperature-dependent constant and μ is the shear modulus at temperature. c_3 is a constant consisting out of the activation energy for diffusional creep (Q_c) and the gas constant (R):

$$c_3 = \frac{Q_c}{R} = \frac{410000}{8.3145} = 49311.4$$
 [in K] . (D2T)

Combining equations (D2Q), (D2R) and (D2S) gives the final equation for $\epsilon_{c,s}(t)$:

$$\epsilon_{c,s}(t) = \epsilon_0 - \left((1-n) \left(\frac{\epsilon_0^{(1-n)}}{1-n} - B_c \cdot \left(\frac{E}{\mu} \right)^n \cdot e^{\frac{-49311.4}{T_K}} \cdot t \right) \right)^{\frac{1}{1-n}}, \tag{D2U}$$

With $T_{\boldsymbol{K}}$ being the temperature in Kelvin.

D3. Mendelson-Roberts-Manson correlation for creep rupture

Although the Larson-Miller method is most frequently used and considered for this research, also the Mendelson-Roberts-Manson (M-R-M) correlation may be used [D1,D2] to determine the minimum time to rupture $(t_{R,min})$ at a certain temperature and stress [D2]:

$$\log_{10}(t_{R,min}) = -20.66 + 37531 \left(\frac{1}{T_K}\right) + 1.20 \log_{10}(\sigma) - \frac{7568}{T_K} \log_{10}(\sigma) ,$$
(D3A)

with $t_{R,min}$ in hours, σ in MPa and T_K in Kelvin.

The Larson-Miller method is considered in the research instead of the Mendelson-Roberts-Manson method because there is more LMP data available and LMP is the method referred to in the ASME.

D4. Relation between the total strain, elastic strain and plastic strain.

When the change in strain amplitude ($\Delta \varepsilon_a$) is plotted as a function of the change in the number of cycle reversals ($\Delta 2N$) and the axes are swapped, the graph as shown in figure D4.1 is constructed, which is applicable for b_f = -0.12 and c_f = -0.76, so in the temperature range of 538 – 704 °C.



Figure D4.1: Change in number of cycles with change in total strain amplitude (538 – 704 °C).

The polynomial equation for the temperature range 538 – 704 °C for $\Delta \varepsilon_a$ higher than zero (red in figure D4.1):

$$\begin{split} \log(\Delta 2N) &= 0.17495 \cdot (\log(\Delta \varepsilon_a))^4 - 0.67926 \cdot (\log(\Delta \varepsilon_a))^3 + 1.0847 \cdot (\log(\Delta \varepsilon_a))^2 - 2.2456 \cdot \log(\Delta \varepsilon_a) - 1.4713 \cdot 10^{-3} \end{split} \tag{D4A}$$

The polynomial equation for the temperature range 538 – 704 °C for $\Delta \varepsilon_a$ lower than zero (blue in figure D4.1):

$$\log(\Delta 2N) = 42.644 \cdot (\log(\Delta \varepsilon_a))^6 + 113.55 \cdot (\log(\Delta \varepsilon_a))^5 + 103.48 \cdot (\log(\Delta \varepsilon_a))^4 + 35.566 \cdot (\log(\Delta \varepsilon_a))^3 + 7.5522 \cdot (\log(\Delta \varepsilon_a))^2 - 1.9466 \cdot \log(\Delta \varepsilon_a) + 2.4949 \cdot 10^{-3}$$
(D4B)

When the elastic strain part becomes that large in comparison to the plastic strain part, the strain amplitude can be described with the elastic strain part equation only. This is considered when the elastic strain part is 99.9% of the strain amplitude (i.e. plastic strain part is 1,000x smaller than the elastic strain part). This is from a change in strain amplitude ($\Delta \varepsilon_a$) smaller (more negative) than -0.80 and this is indicated by the green line in figure D4.1. This line is known to progress through the point (-0.303,0) of the graph, since at the transition point the elastic strain part is half of the total strain amplitude (i.e. LOG(0.5)=-0.303) and there are no additional cycles in comparison to the transition point. This leaves the equation:

$$\log(\Delta 2N) = -9.0909 \cdot \log(\Delta \varepsilon_a) - 2.7366 \tag{D4C}$$

in which the slope is the reciprocal of the slope of the elastic strain part versus number of cycles to failure curve (i.e. -0.11).

For the temperature range of 704 – 871 °C, the change in number of cycle reversals ($\Delta 2N$) plotted as function of the change in strain amplitude ($\Delta \varepsilon_a$) is shown in figure D4.2, which is applicable for b_f = -0.09 and c_f = -0.76.



Figure D4.2: Change in number of cycles with change in total strain amplitude (704 – 871 °C).

The polynomial equation for the temperature range 704 – 871 °C for $\Delta \varepsilon_a$ higher than zero (red in figure D4.2):

$$\begin{split} \log(\Delta 2N) &= 0.19932 \cdot (\log(\Delta \varepsilon_a))^4 - 0.75929 \cdot (\log(\Delta \varepsilon_a))^3 + 1.1779 \cdot (\log(\Delta \varepsilon_a))^2 - 2.2863 \\ &\cdot \log(\Delta \varepsilon_a) - 1.8081 \cdot 10^{-3} \end{split}$$

The polynomial equation for the temperature range 704 – 871 °C for $\Delta \varepsilon_a$ lower than zero (blue in figure D4.2):

(D4D)

(D4F)

$$log(\Delta 2N) = 67.058 \cdot (log(\Delta \varepsilon_a))^6 + 186.15 \cdot (log(\Delta \varepsilon_a))^5 + 171.15 \cdot (log(\Delta \varepsilon_a))^4 + 58.297 \cdot (log(\Delta \varepsilon_a))^3 + 11.109 \cdot (log(\Delta \varepsilon_a))^2 - 1.8378 \cdot log(\Delta \varepsilon_a) - 3.2569 \cdot 10^{-3} (D4E)$$

The green line is for the situation that the elastic strain part is more than 99.9% of the strain amplitude, which is the case for changes in the strain amplitude of -0.70 and more negative:

$$\log(\Delta 2N) = -11.111 \cdot \log(\Delta \varepsilon_a) - 3.3448$$

in which the slope is the reciprocal of the slope of the elastic strain part versus number of cycles to failure curve in figure D4.2 (i.e. -0.09).

For the temperature range of 871 – 982 °C, the change in number of cycle reversals ($\Delta 2N$) plotted as function of the change in strain amplitude ($\Delta \varepsilon_a$) is shown in figure D4.3, which is applicable for b_f = -0.08 and c_f = -0.76.



Figure D4.3: Change in number of cycles with change in total strain amplitude (871 – 982 °C).

The polynomial equation for the temperature range 871 – 982 °C for $\Delta \varepsilon_a$ higher than zero (red in figure D4.3):

$$\begin{split} \log(\Delta 2N) &= 0.13043 \cdot (\log(\Delta \varepsilon_a))^4 - 0.59670 \cdot (\log(\Delta \varepsilon_a))^3 + 1.0670 \cdot (\log(\Delta \varepsilon_a))^2 - 2.2670 \\ & \cdot \log(\Delta \varepsilon_a) - 3.7390 \cdot 10^{-3} \end{split}$$

The polynomial equation for the temperature range 871 – 982 °C for $\Delta \varepsilon_a$ lower than zero (blue in figure D4.3):

(D4G)

(D4I)

$$log(\Delta 2N) = 34.511 \cdot (log(\Delta \varepsilon_a))^6 + 152.69 \cdot (log(\Delta \varepsilon_a))^5 + 164.38 \cdot (log(\Delta \varepsilon_a))^4 + 57.561 \cdot (log(\Delta \varepsilon_a))^3 + 11.160 \cdot (log(\Delta \varepsilon_a))^2 - 4.3495 \cdot log(\Delta \varepsilon_a) - 3.6345 \cdot 10^{-2}$$
(D4H)

The green line is for the situation that the elastic strain part is more than 99.9% of the strain amplitude, which is the case for changes in the strain amplitude of -0.65 and more negative:

$$\log(\Delta 2N) = -12.500 \cdot \log(\Delta \varepsilon_a) - 3.7629$$

in which the slope is the reciprocal of the slope of the elastic strain part versus number of cycles to failure curve in figure D4.3 (i.e. -0.08).

Because the values of the cycles are taken on a log-scale, the total number of cycles reversals to failure $(2N_f)$ is:

$$2N_f = 2N_t \cdot 10^{\log(\Delta 2N)} = 2N_t \cdot \Delta 2N \quad . \tag{D4J}$$

D5. Equation for the time to rupture

The time to rupture (t_R) has to be determined as one of the criteria in the Toolbox. In paragraph 3.2.4, the final equation for calculating the time to rupture is given as equation 3.2B. In this paragraph, the mathematics for constructing this equation is given. The main equations as described in equations (2.8D) and (2.4B) are repeated here:

$$\sigma = 8.2081 \cdot 10^5 \cdot \exp(-3.7920 \cdot 10^{-4}LMP) , \qquad (D5A) \text{ and}$$
$$LMP = T_K (\log t_R + 20) . \qquad (D5B)$$

First, equation (D5B) is rewritten to express t_R as function of the Larson-Miller Parameter (*LMP*) and the temperature in Kelvin (T_K):

$$t_R = 10^{\left(\frac{LMP}{T_K} - 20\right)}$$
 . (D5C).

The second step is to rearrange equation (D5A) in order to express the *LMP* as function of the stress (σ) with using as constants $A_R = 8.2081 \cdot 10^5$ and $B_R = -3.7920 \cdot 10^{-4}$:

$$LMP = \frac{\ln\left(\frac{\sigma}{A_R}\right)}{B_R}$$
 (D5D)

Combining equations (D5C) and (D5D) gives:

$$t_R = 10^{\left(\frac{LMP}{T_K} - 20\right)} = 10^{\left(\frac{\ln\left(\frac{\sigma}{A_R}\right)}{B_R \cdot T_K} - 20\right)}$$
 (D5E)

Equation (D5E) can be reduced further by using:

$$10^{(x-y)} = \frac{10^x}{10^y}$$
 , (D5F)

$$\ln(x) = 2.3026 \cdot \log(x)$$
 . (D5G)

These equations transform equation (D5E) into:

$$t_{R} = \frac{10^{\left(\frac{2.3026 \cdot \log\left(\frac{\sigma}{A_{R}}\right)}{B_{R} \cdot T_{K}}\right)}}{10^{20}} , \qquad (D5H)$$

and by using:

$$10^{z \cdot \log(x)} = x^z \qquad , \tag{D5I}$$

the final result for t_R is obtained:

$$t_R = \frac{\left(\frac{\sigma}{A_R}\right)^{\frac{2.3026}{B_R \cdot T_k}}}{10^{20}}$$
 (D5J)

Filling out A_R and B_R , leads to:

$$t_R = \frac{\left(\frac{\sigma}{8.2081 \cdot 10^5}\right)^{-3.7920 \cdot 10^{-4}T_K}}{10^{20}} = \frac{\left(\frac{\sigma}{8.2081 \cdot 10^5}\right)^{-6.0723 \cdot 10^3}}{T_K}}{10^{20}} \qquad . \tag{D5K}$$

D6. Literature

- [D1] D.K. Fork, J. Fitch, S. Ziaei, R.I. Jetter, *Life Estimation of Pressurized-Air Solar-Thermal Receiver tubes,* Journal of Solar Energy Engineering, SOL-12-1058, 2012.
- [D2] D.R. Eno, G.A. Young, T.-L. Sham, A unified view of engineering creep parameters, Proceedings of PVP 2008 ASME Pressure Vessels and Piping Division Conference, art. 61129, 2008.

Appendix E – Bree diagram

The Bree diagram may be used to investigate the change of ratchetting or plastic shakedown (i.e. lowcycle fatigue) of a component subject to a combined constant primary stress (σ_p) and thermal stress (σ_t), which is caused by a uniform through-wall temperature gradient. This temperature gradient causes the warmer side of the wall to expand more than the colder side and because the colder side of the wall restrains the warmer side from expanding, a bending stress arises. This thermal bending stress is cyclic when temperature variations are present and this stress is described by:

$$\sigma_t = E \cdot \alpha \cdot \Delta T \qquad , \qquad (E1)$$

in which *E* is the elastic modulus, α is the coefficient of thermal expansion and ΔT is the temperature difference.

In case the temperature gradient is considered linear over the wall thickness (s), the thermal stress is always zero at the centre of the wall (i.e. x = 0). For that reason, it is convenient to choose the centre of the wall as zero in the through-wall coordinate system. The surfaces are at:

$$x = \pm \frac{s}{2}$$
 (E2)

The thermal stress at the warmer side of the wall is a compressive stress (i.e. negative stress) due to the constraint, while at the colder side the thermal stress is in tension. The thermal stress progression through the wall ($\sigma_t(x)$) can for that reason be described by:

$$\sigma_t(x) = \frac{2x}{s}\sigma_t$$
 for $-\frac{s}{2} \le x \le \frac{s}{2}$, (E3)

where $-\frac{s}{2}$ is the warmer side of the wall and $\frac{s}{2}$ is the colder side. Although a tension and compression stress exist, the nett thermal stress (σ_t) is the stress of evaluation.

In case the thermal stress is a cycling stress, several cycling modes may occur. These cycling modes and stress combinations are displayed in the Bree diagram in figure E1.1. In the Bree diagram, the constant primary stress (σ_p) and the thermal cyclic stress (σ_t) are given as a normalised stress in relation to the yield strength (σ_y). The primary stress is maximised to the yield strength to avoid yielding. The thermal cyclic stress (σ_t) can be higher than the yield strength, and combined with the primary stress it leads to one of the cyclic modes, which are shown in figure E1.1 as:

- E Elastic cycling.
- S Elastic shakedown.
- P Plastic shakedown (i.e. low-cycle fatigue).
- R Ratchetting.

In the following paragraphs, elastic cycling, elastic shakedown, plastic shakedown and ratchetting are described, including the stress combinations for which these cycling modes appear.



Figure E1.1: Bree diagram for elastic cycling (E), elastic shakedown (S), plastic shakedown (P; i.e. low-cycle fatigue) and ratchetting (R) [E1].

E1. Elastic cycling

Elastic cycling occurs when the combined constant stress and cyclic thermal stress remains always lower than or equal to the yield strength (σ_y) over the entire wall, regardless if both stresses are in tensile or in compression direction:

$$\left|\sigma_{p} + \sigma_{t}\right| \le \left|\sigma_{y}\right|$$

(E4)

This region is shown in the Bree diagram of figure E1.1 by the letter "E".

E2. Yielding

In case some yielding occurs due to the thermal stress (i.e. $\sigma_p + \sigma_t > \sigma_y$). The total stress consists of an elastic part ($\sigma_{el} = \sigma_y$) and a plastic part ($\sigma_{pl} = \sigma_1$). When this stress combination acts on the wall, it may be described as:

$$\frac{\sigma_y + \sigma_1}{\frac{s}{2} - x} = \frac{2\sigma_t}{s} \tag{E5}$$

Now σ_1 can be expressed as equation of the other terms:

$$\sigma_1 = \left(\frac{s}{2} - x\right) \cdot \frac{2\sigma_t}{s} - \sigma_y = \left(1 - \frac{2x}{s}\right)\sigma_t - \sigma_y \qquad . \tag{E6}$$

Because primary stress (σ_p) is equal to or lower than the yield strength (σ_y), any occurring yielding is caused by the thermal stress. Therefore, an equilibrium stress equation for the wall thickness can be constructed in which both primary stress and yield strength are constant over the wall and the local thermal stress depends on the distance from the centre of the wall as described by equation (E3). This equilibrium equation may be given as:

$$s(\sigma_y - \sigma_p) = \frac{1}{2} \left(\frac{s}{2} - x \right) \left(\sigma_y + \sigma_1 \right) \qquad , \tag{E7}$$

in which the constant stresses are given on the left side of the equation and the thermal stress including yielding is given at the right side of the equation. The plastic strain (σ_1) is then:

$$\sigma_1 = \frac{2s(\sigma_y - \sigma_p)}{\binom{s}{2} - x} - \sigma_y = \frac{4(\sigma_y - \sigma_p)}{\left(1 - \frac{2x}{s}\right)} - \sigma_y \qquad (E8)$$

Equations (E6) and (E8) should give the same result, thus:

$$\left(1 - \frac{2x}{s}\right)\sigma_t - \sigma_y = \frac{4(\sigma_y - \sigma_p)}{\left(1 - \frac{2x}{s}\right)} - \sigma_y \qquad , \tag{E9}$$

and to be able to find a location in the wall where a certain total stress is present, equation (E9) is solved for x:

$$\left(1 - \frac{2x}{s}\right)\sigma_t = \frac{4(\sigma_y - \sigma_p)}{\left(1 - \frac{2x}{s}\right)} \to \left(1 - \frac{2x}{s}\right)^2 = \frac{4(\sigma_y - \sigma_p)}{\sigma_t} \quad , \tag{E10}$$

which leads to:

$$x = \frac{s}{2} \left(1 - \sqrt{\frac{4(\sigma_y - \sigma_p)}{\sigma_t}} \right) = \frac{s}{2} \left(1 - 2\sqrt{\frac{(\sigma_y - \sigma_p)}{\sigma_t}} \right)$$
(E11)

The stress part causing plastic strain (σ_1) can then be calculated via equation (E6):

$$\sigma_{1} = \left(1 - \frac{2x}{s}\right)\sigma_{t} - \sigma_{y} = \left(1 - \frac{2}{s} \cdot \frac{s}{2}\left(1 - 2\sqrt{\frac{(\sigma_{y} - \sigma_{p})}{\sigma_{t}}}\right)\right)\sigma_{t} - \sigma_{y} = \left(1 - 1 + 2\sqrt{\frac{(\sigma_{y} - \sigma_{p})}{\sigma_{t}}}\right)\sigma_{t} - \sigma_{y} = 2\sqrt{\frac{(\sigma_{y} - \sigma_{p})}{\sigma_{t}}}\sigma_{t} - \sigma_{y}$$
(E12)

This stress is used later in this chapter to distinguish between shakedown and ratchetting.

E3. Ratchetting - Situation 1: one wall in tensile yield

As described before, thermal stress is caused by a temperature gradient and is therefore maximum at the surfaces and zero at the wall centre in case no constant primary stress is present. Often, both primary constant stress and thermal stress are present and for that reason the compressive stress at the warm wall surface is not equal to the tensile stress at the cold wall. Ratchetting occurs when during one half-cycle more than half of the wall thickness is plastically deformed. This is mathematically explained in the following sub-paragraphs.

E3.1 First half-cycle

In the first half-cycle both constant primary stress and thermal stress are present. An example of a possible stress distribution (σ) and plastic strain distribution (ε_p^1) for a primary stress causing a tensile stress over the wall is shown in figure E1.2, where the plastic strain (ε_p) is still zero at x = -a. This situation causes a plastic strain over more than half of the wall.



Figure E1.2: Stress and strain distribution curves of the first half-cycle.

Regarding equations (E5) and (E6), at x = -a the following strain distribution is found during the first half-cycle:

$$E\varepsilon^1 = \sigma_y + \frac{2a}{s}\sigma_t$$
 at $x = -a$, (E13)

in which ε^1 is the total strain developed during the first half-cycle. Further from the wall surface at the left side, so at x > -a, plastic strain (ε_p^1) becomes part of the total strain:

$$E\varepsilon^1 = \sigma_y + E\varepsilon_p^1 - \frac{2x}{s}\sigma_t$$
 at $x > -a$. (E14)

Combining equations (E13) and (E14) results in an equation for the plastic strain development after the first half-cycle:

$$\varepsilon_p^1 = 2\frac{\sigma_t}{E} \left(\frac{a}{s} + \frac{x}{s}\right) \tag{E15}$$

The maximum plastic strain ($\varepsilon_{p,max}^1$) is found at the wall surface at the right side (i.e. at $x = \frac{s}{2}$):

$$\varepsilon_{p,max}^{1} = \frac{2\sigma_{t}}{E} \left(\frac{a}{s} + \frac{s}{2s} \right) = \frac{\sigma_{t}}{E} \left(\frac{2a}{s} + 1 \right) \quad , \tag{E16}$$

which is indicated by "1" in figure E1.2.

E3.2 Second half-cycle

During the second half-cycle, the thermal strain is removed (i.e. no temperature gradient over the wall) and as a result the following equation applies:

$$E\varepsilon^{2} = \sigma + E\varepsilon_{p}^{1+2} \to \sigma = E\varepsilon^{2} - E\varepsilon_{p}^{1+2} \qquad . \tag{E17}$$

This equation is equation (E14), but then without the thermal stress. Equation (E17) requires the stress distribution to develop in the equal opposite direction as the plastic strain distribution. The plastic strain distribution is given by equation (E15), so the stress distribution for the second half-cycle is:

$$\sigma = \sigma_y + 2\sigma_t \left(\frac{a}{s} - \frac{x}{s}\right)$$
 for $x > a$, (E18)

and $\sigma = \sigma_y$ for $x \le a$. This stress distribution is shown in figure E1.3. Because the slope of the stress distribution curve is non-zero, no yielding occurs, so for x > a the $\varepsilon_p^2 = 0$, and:

$$\varepsilon_p^{1+2} = \varepsilon_p^1 = 2\frac{\sigma_t}{E} \left(\frac{a}{s} + \frac{x}{s}\right) \qquad \text{for} \qquad x > a \quad . \tag{E19}$$

For the second half-cycle for x > a, equations (E17), (E18) and (E19) are combined into:

$$E\varepsilon^{2} = \sigma + E\varepsilon_{p}^{1} = \sigma_{y} + 2\sigma_{t}\left(\frac{a}{s} - \frac{x}{s}\right) + 2\frac{\sigma_{t} \cdot E}{E}\left(\frac{a}{s} + \frac{x}{s}\right) = \sigma_{y} + 4\frac{a}{s}\sigma_{t}$$
(E20)

The stress distribution curve is $\sigma = \sigma_y$ for $x \le a$, resulting in no slope of this curve and hence yielding takes place (i.e. $\varepsilon_p^2 \ne 0$). For this part of the wall, the combination of equations (E17) and (E20) and $\sigma = \sigma_y$ leads to a strain distribution of:

$$E\varepsilon_p^{1+2} = 4\frac{a}{s}\sigma_t \to \varepsilon_p^{1+2} = \frac{4a}{s} \cdot \frac{\sigma_t}{E} \qquad \text{for } x \le a \qquad , \tag{E21}$$

which is shown in figure E1.3 as well.



Figure E1.3: Stress and strain distribution curves of the second half-cycle.

E3.3 Third half-cycle

The stress distribution curve for the third half-cycle is identical to that of the first half-cycle:

$$\sigma = \sigma_y + 2\sigma_t \left(\frac{a}{s} + \frac{x}{s}\right) \qquad \text{for} \qquad x < -a \qquad , \tag{E22}$$

and $\sigma = \sigma_y$ for $x \ge a$, which is shown again in figure E1.4. In the wall at the locations x < -a, the stress curve slope is non-zero, hence additional yielding does not occur and the plastic strain remains unchanged in comparison to the plastic strain accumulated after the second half-cycle (i.e. equation (E21) applies to the wall locations x < -a). This is shown in the strain distribution curve of figure E1.4. The strain distribution for x < -a is a combination of the stress distribution equation (E22), the strain distribution of the second half-cycle (E21) and the thermal stress progression equation (E3):

$$E\varepsilon^{3} = \sigma_{y} + 2\sigma_{t}\left(\frac{a}{s} + \frac{x}{s}\right) + E\varepsilon_{p}^{1+2} - \frac{2x}{s}\sigma_{t} \qquad \text{for} \qquad x < -a.$$
(E23)

Using equation (E21) to substitute $E\varepsilon_p^{1+2}$, leads to:

$$E\varepsilon^{3} = \sigma_{y} + \frac{\sigma_{t}}{s} \left((2a + 2x) + 4a - 2x \right) = \sigma_{y} + \frac{6a}{s} \sigma_{t} \quad \text{for } x < -a.$$
 (E24)

As a check, the plastic strain in the third half-cycle $(E\varepsilon_p^3)$ for x < -a is obtained by combined equations (E13) and (E24):

$$E\varepsilon_p^3 = E\varepsilon^3 - E\varepsilon^1 = \left(\sigma_y + \frac{6a}{s}\sigma_t\right) - \left(\sigma_y + \frac{2a}{s}\sigma_t\right) = \frac{4a}{s}\sigma_t \to \varepsilon_p^3 = \frac{4a}{s} \cdot \frac{\sigma_t}{E} = \varepsilon_p^{1+2}$$

for $x < -a$. (E25)

Further from the wall surface at the left side, so at $x \ge -a$, plastic strain strain accumulated during the first three cycles (ε_p^{1+2+3}) becomes part of the total strain (see also equation (E14)):

$$E\varepsilon^3 = \sigma_y + E\varepsilon_p^{1+2+3} - \frac{2x}{s}\sigma_t$$
 at $x \ge -a$, (E26)

Substitution of equation (E24) in equation (E26) leads to:

$$\sigma_y + \frac{6a}{s}\sigma_t = \sigma_y + E\varepsilon_p^{1+2+3} - \frac{2x}{s}\sigma_t \to \varepsilon_p^{1+2+3} = \frac{\sigma_t}{E}\left(\frac{6a}{s} + \frac{2x}{s}\right) = \frac{2\sigma_t}{E}\left(\frac{3a}{s} + \frac{x}{s}\right)$$

for $x \ge -a$. (E27)

The maximum plastic strain ($\varepsilon_{p,max}^{1+2+3}$) after the third half-cycle is found at the colder wall surface (i.e. at $x = \frac{s}{2}$), which is indicated by "2" in figure E1.4. This maximum plastic may be equated as:



Figure E1.4: Stress and strain distribution curves of the third (left) and fourth (right) half-cycle.

E3.4 Fourth half-cycle

Just as in the second half-cycle, the thermal strain is removed in the fourth half-cycle and equation (E17) can be changed into:

$$E\varepsilon^4 = \sigma + E\varepsilon_p^{1+2+3+4} \to \sigma = E\varepsilon^2 - E\varepsilon_p^{1+2+3+4}$$
(E29)

The stress distribution for the fourth half-cycle is the same as that for the second half-cycle:

$$\sigma = \sigma_y + 2\sigma_t \left(\frac{a}{s} - \frac{x}{s}\right) \qquad \text{for} \qquad x > a \quad , \tag{E30}$$

and $\sigma = \sigma_y$ for $x \le a$. This stress distribution is shown in figure E1.4.

Because the slope of the stress distribution curve is non-zero, no yielding occurs, which means that for x > a the $\varepsilon_p^4 = 0$, so:

$$\varepsilon_p^{1+2+3+4} = \varepsilon_p^{1+2+3} = 2\frac{\sigma_t}{E} \left(\frac{3a}{s} + \frac{x}{s}\right) \quad \text{for} \quad x > a \quad .$$
 (E31)

In the fourth half-cycle for x > a, equations (E29), (E30) and (E31) are combined into:

$$E\varepsilon^{4} = \sigma + E\varepsilon_{p}^{1+2+3} = \sigma_{y} + 2\sigma_{t}\left(\frac{a}{s} - \frac{x}{s}\right) + 2\frac{\sigma_{t} \cdot E}{E}\left(\frac{3a}{s} + \frac{x}{s}\right) = \sigma_{y} + 8\frac{a}{s}\sigma_{t}$$
for $x > a$. (E32)

For $x \le a$, the stress distribution curve is $\sigma = \sigma_y$ and $\varepsilon_p^4 \ne 0$. For this part of the wall, the combination of equations (E29) and (E32) and $\sigma = \sigma_y$ leads to a strain distribution of:

$$E\varepsilon_p^{1+2+3+4} = 8\frac{a}{s}\sigma_t \to \varepsilon_p^{1+2+3+4} = \frac{8a}{s} \cdot \frac{\sigma_t}{E} \qquad \text{for } x \le a \qquad , \tag{E33}$$

E3.5 Subsequent half-cycles

When comparing the loading half-cycles (i.e. first and third half-cycle, so with thermal strain), the plastic strain distribution curve is shifted upwards from the first half-cycle to the third half-cycle ($\Delta \varepsilon_p^{1\rightarrow 3}$) with an amount of:

$$\Delta \varepsilon_p^{1 \to 3} = \frac{4a}{s} \cdot \frac{\sigma_t}{E} \tag{E34}$$

For the unloading half-cycles (i.e. second and fourth half-cycle, so without thermal strain), the same shift in the plastic strain distribution curve ($\Delta \varepsilon_p^{2\rightarrow 4}$) is observed:

$$\Delta \varepsilon_p^{2 \to 4} = \frac{4a}{s} \cdot \frac{\sigma_t}{E} \tag{E35}$$

This shift is continued during each subsequent cycle, so the wall continues to strain from one cycle to the next cycle. This process is called ratchetting and the ratchetting strain per cycle (ε_{rat}) is:

$$\varepsilon_{ratch.} = \frac{4a}{s} \cdot \frac{\sigma_t}{E}$$
 (E36)

E4. Elastic shakedown

Ratchetting is avoided in case cyclic strain does not accumulate, which is achieved when a = x = 0.

E4.1 First half-cycle

Regarding equations (E13) and (E14), at a = x = 0, the following strain distribution is found during the first half-cycle:

$$E\varepsilon^{1} = \sigma_{y} + \frac{2x}{s}\sigma_{t} \qquad \text{for} \qquad x \le 0 \qquad , \qquad (E37)$$
$$E\varepsilon^{1} = \sigma_{y} + E\varepsilon^{1}_{p} - \frac{2x}{s}\sigma_{t} \qquad \text{for} \qquad x > 0 \qquad . \qquad (E38)$$

The stress and strain distributions of this situation are given in the left figure of figure E1.5.

Combining equations (E37) and (E38) results in an equation for the plastic strain development after the first half-cycle:

$$\varepsilon_p^1 = 2 \frac{\sigma_t}{E} \left(\frac{x}{s} \right)$$
 for $x > 0$. (E39)

The maximum plastic strain ($\varepsilon_{p,max}^1$) is found at the wall surface at the right side (i.e. at $x = \frac{s}{2}$):

$$\varepsilon_{p,max}^{1} = \frac{2\sigma_{t}}{E} \left(\frac{s}{2s}\right) = \frac{\sigma_{t}}{E} \qquad , \tag{E40}$$

which is indicated by "1" in figure E1.5.

E4.2 Second half-cycle

For the second half-cycle, the thermal strain is removed and equation (E17) applies:

$$E\varepsilon^{2} = \sigma + E\varepsilon_{p}^{1+2} \to \sigma = E\varepsilon^{2} - E\varepsilon_{p}^{1+2} \qquad . \tag{E41}$$

Equation (E41) requires the stress distribution to develop in the equal opposite direction as the plastic strain distribution. The plastic strain distribution is given by equation (E39), so the stress distribution for the second half-cycle may be described as:

$$\sigma = \sigma_y + 2\sigma_t \left(-\frac{x}{s}\right) = \sigma_y - 2\sigma_t \left(\frac{x}{s}\right) \qquad \text{for} \qquad x > 0 \quad , \tag{E42}$$

and $\sigma=\sigma_y$ for $x\leq 0$. This stress distribution is shown in figure E1.5.

There is no yielding occurring in the wall at locations x > 0 (i.e. $\varepsilon_p^2 = 0$), so

$$\varepsilon_p^{1+2} = \varepsilon_p^1 = 2\frac{\sigma_t}{E} \left(\frac{x}{s}\right) \quad \text{for} \quad x > 0 \quad .$$
 (E43)

For the second half-cycle, for x > 0, equations (E41), (E42) and (E43) are combined into:

$$E\varepsilon^{2} = \sigma + E\varepsilon_{p}^{1} = \sigma_{y} - 2\sigma_{t}\left(\frac{x}{s}\right) + 2\frac{\sigma_{t} \cdot E}{E}\left(\frac{x}{s}\right) = \sigma_{y} \qquad (E44)$$

The stress distribution curve is $\sigma = \sigma_y$ for $x \le 0$, plastic strain ε_p^2 may develop. The combination of equations (E41) and (E44) and $\sigma = \sigma_y$ leads to a strain distribution of:

$$E\varepsilon_p^{1+2} = E\varepsilon_p^1 \to \varepsilon_p^2 = 0$$
 for $x \le 0$, (E45)

which is shown in the right figure of figure E1.5 as well. This means that no plastic strain develops at all during the second half-cycle.



Figure E1.5: Stress and strain distribution curves of the first half-cycle in case a = x = 0 (left) and of the second half-cycle (right).

E4.3 Third half-cycle

The third half-cycle shall following the same stress distribution path and strain distribution path as the first half-cycle. This means that:

$$\varepsilon_p^{1+2+3} = \varepsilon_p^1 + \varepsilon_p^2 + \varepsilon_p^3 = \varepsilon_p^1 \qquad (E46)$$

From the second half-cycle evaluation it is already known that $\varepsilon_p^2 = 0$ for all values of x. From equation (E46) follows that $\varepsilon_p^3 = 0$ for all values of x as well.

E4.4 Fourth half-cycle

The fourth half-cycle follows the same path as the second half-cycle and since $\varepsilon_p^2 = \varepsilon_p^3 = 0$ for all values of x, the result is:

$$\varepsilon_p^{1+2+3+4} = \varepsilon_p^{1+2} \to \varepsilon_p^{1+4} = \varepsilon_p^1 \qquad (E47)$$

Equation (E47) can only be valid when $\varepsilon_p^4 = 0$.

This means that plastic strain develops in the first half-cycle only and all subsequent half-cycles are causing elastic strains only. This process is called elastic shakedown.

E4.5 Boundary in Bree diagram between R1 and S1

In the Bree diagram, the boundary for ratchetting is found by using equation (E11) where a = x = 0:

$$x = \frac{s}{2} \left(1 - \sqrt{\frac{4(\sigma_y - \sigma_p)}{\sigma_t}} \right) = \frac{s}{2} \left(1 - 2\sqrt{\frac{(\sigma_y - \sigma_p)}{\sigma_t}} \right) = 0$$
 (E46)

This requirement can only be met when:

$$1 - 2\sqrt{\frac{(\sigma_y - \sigma_p)}{\sigma_t}} = 0 \to 2\sqrt{\frac{(\sigma_y - \sigma_p)}{\sigma_t}} = 1 \to \sqrt{\frac{(\sigma_y - \sigma_p)}{\sigma_t}} = \frac{1}{2} \qquad , \tag{E47}$$

which is squared at both sides and leads to:

$$\frac{(\sigma_y - \sigma_p)}{\sigma_t} = \frac{1}{4} \to (\sigma_y - \sigma_p) = \frac{\sigma_t}{4} \to \sigma_p + \frac{\sigma_t}{4} = \sigma_y \qquad (E48)$$

Equation (E48) defines the boundary between ratchetting (R₁) and elastic shakedown (S₁) as shown in figure E1.1.

E5. Ratchetting – Situation 2: one wall in tensile yield, other in compressive yield

In case the thermal strain is high (i.e. large temperature gradient) and the constant primary stress is relatively low, the situation may occur that one of the surfaces is subject to a compressive stress that causes yielding, while the other surface undergoes a tensile stress that causes yielding as well. Such a stress distribution is shown in figure E1.6, in which the material between -a and -b does not yield (i.e. elastic strain only) and the constant primary stress causes a tensile stress on the wall.

E5.1 First half-cycle

The plastic strain distribution is caused by both primary stress (σ_p) and thermal stress (σ_t) which is shown in figure E1.6 and the equilibrium equation becomes:

$$\sigma_p \cdot s = \sigma_y \left(\frac{s}{2} - (-a)\right) - \left(-\sigma_y\right) \left(\frac{-s}{2} - (-b)\right) = \sigma_y \left(\frac{s}{2} + a\right) - \sigma_y \left(\frac{s}{2} - b\right) = \sigma_y (a+b)$$
(E49)



Figure E1.6: Stress and strain distribution curves of the first half-cycle.

The plastic strain is zero at -b < x < -a and the stress expands from $-\sigma_y$ at -b to σ_y at -a, thus:

$$\sigma(x) = -\sigma_y + 2\sigma_y \left(\frac{x + (0 - (-b))}{((0 - (-b) - (0 - (-a))))}\right) = -\sigma_y + 2\sigma_y \left(\frac{x + b}{b - a}\right) \quad .$$
(E50)

The stress distribution includes the thermal stress variation as well:

$$E\varepsilon = \sigma(x) - \frac{2x}{s}\sigma_t = -\sigma_y + 2\sigma_y \left(\frac{x+b}{b-a}\right) - \frac{2x}{s}\sigma_t \qquad . \tag{E51}$$

The mathematics are made less cumbersome by expressed the constant primary stress as a ratio of the yield strength:

$$X = \frac{\sigma_p}{\sigma_y} \qquad , \tag{E52}$$

and the thermal stress as a ratio of the yield strength:

$$Y = \frac{\sigma_t}{\sigma_y} \tag{E53}$$

Equation (E49) can therefore be changed to:

$$a+b = \frac{\sigma_p}{\sigma_y}s = X \cdot s \qquad , \tag{E54}$$

and equation (E51) becomes:

$$E\varepsilon = -\sigma_y + 2\sigma_y \left(\frac{x+b}{b-a}\right) - \frac{2x}{s} Y \sigma_y \qquad . \tag{E55}$$

The dependency of *x* in equation (E55) is eliminated by:

$$\left(\frac{x}{b-a}\right) = \frac{x}{s}Y \to b - a = \frac{s}{Y} \qquad , \tag{E56}$$

which leads to a renewed equation for equation (E55):

$$E\varepsilon = -\sigma_y + 2\sigma_y \left(\frac{b}{b-a}\right) \tag{E57}$$

Now a can be calculated by using equations (E54) and (E56):

$$(a+b) - (b-a) = 2a = X \cdot s - \frac{s}{Y} = s\left(X - \frac{1}{Y}\right) \to a = \frac{1}{2}s\left(X - \frac{1}{Y}\right),$$
 (E58)

and *b* can be calculated by:

$$(a+b) + (b-a) = 2b = X \cdot s + \frac{s}{Y} = s\left(X + \frac{1}{Y}\right) \to b = \frac{1}{2}s\left(X + \frac{1}{Y}\right).$$
 (E59)

Substituting the a and b in equation (E57) results into:

$$E\varepsilon = -\sigma_y + 2\sigma_y \left(\frac{\frac{1}{2}s\left(x + \frac{1}{Y}\right)}{\frac{1}{2}s\left(\left(x + \frac{1}{Y}\right) - \left(x - \frac{1}{Y}\right)\right)}\right) = -\sigma_y + 2\sigma_y \left(\frac{\left(x + \frac{1}{Y}\right)}{\frac{2}{Y}}\right) = -\sigma_y + \sigma_y (XY + 1) = -\sigma_y + XY\sigma_y + \sigma_y = XY\sigma_y \quad \text{for } -b < x < -a. \quad (E60)$$

For x < -b, plastic strain is present and the following stress distribution equation can be constructed:

$$E\varepsilon = E\varepsilon_p^1 - \sigma_y - \frac{2x}{s}Y\sigma_y = XY \quad , (E61)$$

which leads to ε_p^1 :

$$\varepsilon_p^1 = \frac{\sigma_y}{E} \left(XY + \frac{2x}{s}Y + 1 \right) \tag{E62}$$

At x = -b, the plastic strain is still zero and at $x = \frac{-s}{2}$, the maximum plastic strain is found:

$$\varepsilon_{p,max}^{1}\left(x=\frac{-s}{2}\right) = \frac{\sigma_{y}}{E}(XY-Y+1) \qquad , \tag{E63}$$

which is indicated by "1" in figure E1.6.

Also for x > -a, plastic strain is present and the stress distribution equation is:

$$E\varepsilon = E\varepsilon_p^1 + \sigma_y - \frac{2x}{s}Y\sigma_y = XY\sigma_y \qquad . \tag{E64}$$

which leads to ε_p^1 :

$$\varepsilon_p^1 = \frac{\sigma_y}{E} \left(XY + \frac{2x}{s}Y - 1 \right) \tag{E65}$$

At x = -a, the plastic strain is still zero and at $x = \frac{s}{2}$, the maximum plastic strain is found:

$$\varepsilon_{p,max}^{1}\left(x=\frac{s}{2}\right) = \frac{\sigma_{y}}{E}(XY+Y-1) \qquad , \tag{E66}$$

which is indicated by "2" in figure E1.6.

E5.2 Second half-cycle

The thermal strain is removed in this half-cycle. The stress distribution curve of the total strain is independent of x and therefore the curve of the second half-cycle runs equal and opposite to the $E\varepsilon_p^1$ curve. This results in the stress distribution curve of the second half-cycle as shown in figure E1.7. First, the area without plastic strain caused by the second half-cycle (ε_p^2) is considered, i.e. a < x < b. In this part, the plastic strain of the first half-cycle is present, which is given by equation (E65) and the strain distribution equation without the thermal strain.

$$E\varepsilon = E\varepsilon_p^1 - \sigma(x) = \left(XY + \frac{2x}{s}Y - 1\right)\sigma_y + \sigma_y - 2\sigma_y\left(\frac{x + (0-a)}{((0-(-b)-(0-(-a))))}\right) = \left(XY + \frac{2x}{s}Y\right)\sigma_y - 2\sigma_y\left(\frac{x-a}{(b-a)}\right)$$
for $a < x < b$. (E67)

The dependency of x is removed by incorporating equation (E56):



Figure E1.7: Stress and strain distribution curves of the second half-cycle.

Using equation (E58) and (E59) for substituting a and b leads to:

$$E\varepsilon = \sigma_{y}\left(XY + 2\left(\frac{\frac{1}{2}s\left(X - \frac{1}{Y}\right)}{\frac{1}{2}s\left(\left(X + \frac{1}{Y}\right) - \left(X - \frac{1}{Y}\right)\right)}\right)\right) = \sigma_{y}\left(XY + 2\left(\frac{\left(X - \frac{1}{Y}\right)}{\frac{2}{Y}}\right)\right) = \sigma_{y}\left(XY + Y\left(X - \frac{1}{Y}\right)\right) = \sigma_{y}\left(2XY - 1\right) \qquad \text{for } a < x < b \ . \tag{E69}$$

For locations x < a, the strain distribution is:

$$E\varepsilon = E\varepsilon_p^{1+2} - \sigma(x) = E\varepsilon_p^{1+2} + \sigma_y = \sigma_y(2XY - 1)$$
(E70)

Resulting in a plastic strain after the second half-cycle ($E\varepsilon_p^{1+2}$) of:

$$\varepsilon_p^{1+2} = \frac{\sigma_y}{E} (2XY - 1) - \sigma_y = 2\frac{\sigma_y}{E} (XY - 1)$$
 for $x < a$, (E71)

which is independent of x and shown in figure E1.7 by "1".

For x > b, the strain distribution is:

$$E\varepsilon = E\varepsilon_p^{1+2} + \sigma_y = \sigma_y(2XY - 1)$$
(E72)

Resulting in a plastic strain after the second half-cycle ($E\varepsilon_p^{1+2}$) of:

$$\varepsilon_p^{1+2} = \frac{\sigma_y}{E} (2XY - 1) + \frac{\sigma_y}{E} = 2\frac{\sigma_y}{E} XY \text{ for } x > b$$
, (E73)

which is independent of x and indicated by "2" in figure E1.7.

Although the strain distribution equation in the current form does not reveal whether the plastic strains are positive or negative, the strain distribution of the plastic strain is drawn in figure E1.7 as positive values over the entire wall thickness. This can be explained because it is known that during the first half-cycle a plastic strain with a positive value developed in the region a < x < b (see figure E1.6) and no plastic strain developed in this region during the second half-cycle. At the same time, in the region -b < x < -a, there was no plastic strain present in the first half-cycle, while at the second half-cycle a positive stress was active, hence a positive plastic strain has developed.

E5.3 Third half-cycle

In the third half-cycle, the stress distribution is similar to that of the first half-cycle. For the strain distribution, equation (E51) is re-introduced for -b < x < -a, though now including the additional plastic strain accumulated during the first and second half-cycles and knowing that no additional plastic strain develops in this region during the third half-cycle (i.e. $\varepsilon_p^3 = 0$).

$$E\varepsilon = E\varepsilon_p^{1+2} + \sigma(x) - \frac{2x}{s}\sigma_t \quad \text{for} \quad -b < x < -a \quad , \quad (E74)$$

where ε_p^{1+2} evolves from equation (E71)(because both – a and -b are smaller than a), $\sigma(x)$ comes from equation (E50), and the conversion between σ_t and σ_v is done via equation (E53):

$$E\varepsilon = 2\sigma_y(XY - 1) - \sigma_y + 2\sigma_y\left(\frac{x+b}{b-a}\right) - \frac{2x}{s}Y\sigma_y \qquad . \tag{E75}$$

Removal of the x dependency via equation (E56), results into:

$$E\varepsilon = \sigma_{y}\left(2(XY-1) - 1 + 2\left(\frac{b}{b-a}\right)\right) = \sigma_{y}\left(2XY - 3 + 2\left(\frac{b}{b-a}\right)\right).$$
(E76)

Using equation (E58) and (E59) for substituting a and b, leads to:

$$E\varepsilon = \sigma_y \left(2XY - 3 + 2\frac{\frac{1}{2}s\left(x + \frac{1}{Y}\right)}{\frac{1}{2}s\left(\left(x + \frac{1}{Y}\right) - \left(x - \frac{1}{Y}\right)\right)} \right) = \sigma_y \left(2XY - 3 + 2\left(\left(\frac{\left(x + \frac{1}{Y}\right)}{\frac{2}{Y}}\right)\right) \right) = \sigma_y (3XY - 2)$$

for $-b < x < -a$. (E77)

This is shown in figure E1.8 by the horizontal blue line between -b and -a.

For x < -b, additional plastic strain occurs and the stress distribution equation becomes:

$$E\varepsilon = E\varepsilon_p^{1+2+3} - \sigma_y - \frac{2x}{s}Y\sigma_y = \sigma_y(3XY - 2) \quad , \tag{E78}$$

which leads to ε_p^{1+2+3} :

$$\varepsilon_p^{1+2+3} = \frac{\sigma_y}{E} \left(3XY + \frac{2x}{s}Y - 1 \right)$$
 for $x < -b$. (E79)

At x = -b, the additional plastic strain accumulating during the third half-cycle is still zero and at $x = \frac{-s}{2}$, the maximum plastic strain is found:

$$\varepsilon_{p,max}^{1+2+3}\left(x=\frac{-s}{2}\right) = \frac{\sigma_y}{E}(3XY-Y-1)$$
, (E80)

which is indicated by "3" in figure E1.8.

Also for x > -a, additional plastic strain develops as well and the stress distribution equation adherent is:

$$E\varepsilon = E\varepsilon_p^{1+2+3} + \sigma_y - \frac{2x}{s}Y\sigma_y = \sigma_y(3XY - 2) \quad . \tag{E81}$$

which leads to ε_p^{1+2+3} :

$$\varepsilon_p^{1+2+3} = \frac{\sigma_y}{E} \left(3XY + \frac{2x}{s}Y - 3 \right)$$
 for $x > -a$. (E82)

At x = -a, the plastic strain is still zero and at $x = \frac{s}{2}$, the maximum plastic strain is found:

$$\varepsilon_{p,max}^{1+2+3}\left(x=\frac{s}{2}\right) = \frac{\sigma_y}{E}(3XY+Y-3)$$
, (E83)

which is indicated by "4" in figure E1.8.



Figure E1.8: Stress and strain distribution curves of the third half-cycle.

E5.4 Subsequent half-cycles

When comparing the loading half-cycles (i.e. first and third half-cycle, so with thermal strain), the plastic strain distribution curve is shifted upwards from the first half-cycle to the third half-cycle $(\Delta \varepsilon_p^{1\to 3})$ with an amount of (referenced at the right wall, i.e. at $x = \frac{s}{2}$):

$$\Delta \varepsilon_p^{1 \to 3} = \varepsilon_p^3 - \varepsilon_p^1 = \frac{\sigma_y}{E} (3XY + Y - 3) - \frac{\sigma_y}{E} (XY + Y - 1) = \frac{\sigma_y}{E} (2XY - 2) = \frac{2\sigma_y}{E} (XY - 1)$$
(E84)

This shift of the strain distribution curve between the first and the third half-cycle is the same for each point on the strain distribution curve. During each subsequent cycle, the wall continues to strain plastically with the same amount from one cycle to the next cycle (i.e. ratchetting occurs). The ratchetting strain per cycle (ε_{ratc}) is therefore:

$$\varepsilon_{ratc} = \frac{2\sigma_y}{E} (XY - 1)$$
(E85)

E5.5 Boundary in Bree diagram between R₂ and P

Ratchetting does not occur when the ratchetting strain per cycle is zero:

$$\varepsilon_{ratc} = \frac{2\sigma_y}{E}(XY - 1) = 0 \qquad , \tag{E86}$$

which can only be achieved when:

$$XY = 1$$
 . (E87)

Using equations (E52) and (E53) to substitute X and Y results in:

$$XY = \frac{\sigma_p}{\sigma_y} \cdot \frac{\sigma_t}{\sigma_y} = \frac{\sigma_p \cdot \sigma_t}{\sigma_y^2} = 1 \to \sigma_p \cdot \sigma_t = \sigma_y^2 \qquad . \tag{E88}$$

Equation (E88) defines the boundary between ratchetting (R_2) and plastic shakedown (P) as shown in figure E1.1.

E6. Plastic shakedown

Plastic shakedown occurs when cyclic plastic strain develops near the wall surfaces while the centre of the wall is (initially) only exposed to elastic strain. This is best explained by adjusting ratchetting situation 2 and take a instead of -a for one of the plasticity boundaries. The adherent stress distribution curve is shown in figure E1.9.

E6.1 First half-cycle

The equilibrium equation becomes a minor adjustment of equation (E49):

$$\sigma_p \cdot s = \sigma_y \left(\frac{s}{2} - a\right) - \left(-\sigma_y\right) \left(\frac{-s}{2} - (-b)\right) = \sigma_y \left(\frac{s}{2} - a\right) - \sigma_y \left(\frac{s}{2} - b\right) = \sigma_y (b - a)$$
(E89)

Using equation (E52) and eliminating σ_y from the equation, leads to:

$$X \cdot s = b - a \tag{E90}$$

In the wall between -b < x < a, the plastic strain is zero (i.e. $\varepsilon_p^1 = 0$) and equation (E51) becomes:

$$E\varepsilon = -\sigma_y + 2\sigma_y \left(\frac{x+b}{a+b}\right) - \frac{2x}{s}\sigma_t \qquad , \tag{E91}$$

and incorporating equation (E53) as well, results into:

$$E\varepsilon = -\sigma_y + 2\sigma_y \left(\frac{x+b}{a+b}\right) - \frac{2x}{s} Y \sigma_y \qquad . \tag{E92}$$

This equation is made independent of x by:

$$a+b=\frac{s}{\gamma} \qquad , \qquad (E93)$$

which reduces equation (E92) into:

$$E\varepsilon = -\sigma_y + 2\sigma_y \left(\frac{b}{a+b}\right) = \sigma_y \left(2\left(\frac{b}{a+b}\right) - 1\right)$$
(E94)



Figure E1.9: Stress and strain distribution curves of the first half-cycle.

Now a can be calculated by using equations (E90) and (E93):

$$(a+b) - (b-a) = 2a = \frac{s}{Y} - X \cdot s = s\left(\frac{1}{Y} - X\right) \to a = \frac{1}{2}s\left(\frac{1}{Y} - X\right),$$
 (E95)

and *b* can be calculated by:

$$(a+b) + (b-a) = 2b = \frac{s}{Y} + X \cdot s = s\left(\frac{1}{Y} + X\right) \to b = \frac{1}{2}s\left(\frac{1}{Y} + X\right).$$
 (E96)

Substitution of a and b in equation (E94) gives:

$$E\varepsilon = \sigma_{\mathcal{Y}}\left(2\left(\frac{\frac{1}{2}s\left(\frac{1}{Y}+X\right)}{\frac{1}{2}s\left(\left(\frac{1}{Y}-X\right)+\left(\frac{1}{Y}+X\right)\right)}\right) - 1\right) = \sigma_{\mathcal{Y}}\left(\frac{2\left(\frac{1}{Y}+X\right)}{\left(\frac{2}{Y}\right)} - 1\right) = \sigma_{\mathcal{Y}}\left((1+XY) - 1\right) = \sigma_{\mathcal{Y}}XY$$
for $-b < x < a$ (E97)

For x < -b, plastic stain develops and equation (E61) is repeated:

$$E\varepsilon = E\varepsilon_p^1 - \sigma_y - \frac{2x}{s}Y\sigma_y = XY\sigma_y \quad , \tag{E98}$$

which leads to ε_p^1 :

$$\varepsilon_p^1 = \frac{\sigma_y}{E} \left(XY + \frac{2x}{s}Y + 1 \right)$$
(E99)

E-19

At x = -b, the plastic strain is still zero and at $x = \frac{-s}{2}$, the maximum plastic strain is found:

$$\varepsilon_{p,max}^{1}\left(x=\frac{-s}{2}\right)=\frac{\sigma_{y}}{E}(XY-Y+1)$$
 , (E100)

which is identical to the outcome of equation (E63) and indicated by "1" in figure E1.9.

Also for x > a, plastic strain is present and the stress distribution equation is:

$$E\varepsilon = E\varepsilon_p^1 + \sigma_y - \frac{2x}{s}Y\sigma_y = XY\sigma_y \qquad . \tag{E101}$$

which leads to ε_p^1 :

$$\varepsilon_p^1 = \frac{\sigma_y}{E} \left(XY + \frac{2x}{s}Y - 1 \right) \tag{E102}$$

At x = -a, the plastic strain is still zero and at $x = \frac{s}{2}$, the maximum plastic strain is found:

$$\varepsilon_{p,max}^{1}\left(x=\frac{s}{2}\right) = \frac{\sigma_{y}}{E}(XY+Y-1) \qquad , \tag{E103}$$

which is the same value as the result from equation (E66) and indicated by "2" in figure E1.9.

E6.2 Second half-cycle

After removal of the thermal load, the stress distribution curve can be constructed as follows:

1. For −*b* < *x* < *a*.

Because the total strain (ε) is independent on x via equation (E97) and no plastic strain has developed during the first half-cycle (i.e. $\varepsilon_p^1 = 0$), the stress is not dependent on x as well. This means that the stress in this region is constant and between the compressive and tensile yield strength. This stress is same as calculated by equation (E97) (i.e. $E\varepsilon = \sigma_y XY$).

2. Near the surfaces.

The material near the surfaces are still yielding due to the plastic strains developed during the first half-cycle. Since yielding occurs, the stress near the surfaces has a constant value equal to the yield strength.

3. Based on the conclusions from number 1 and 2, there has to be a boundary between the left surface and -b from which the stress starts to change (i.e. a region without additional plastic strain developing during the second half-cycle) until -b is reached. This boundary exists between the right surface and a as well.

Regarding the bullets above, a stress distribution curve may look like shown in figure E1.10, in which the boundaries described in bullet 3 are indicated with -c and c (i.e. assuming these boundaries are situated at the same distance from either wall surface).

Then considered is the region -c < x < -b, where plastic strain has developed during the first halfcycle via equation (E99) and no plastic strain occurs during the second half-cycle. The stress in this region may be equated by:

$$E\varepsilon = \sigma + E\left(\varepsilon_p^1 + \varepsilon_p^2\right) = \sigma + \sigma_y\left(XY + \frac{2x}{s}Y + 1\right)$$
(E104)

F-20

For x = -c, the $\sigma = \sigma_y$, so:

$$E\varepsilon = \sigma_y + \sigma_y \left(XY + \frac{2(-c)}{s}Y + 1 \right) = \sigma_y \left(XY - \frac{2c}{s}Y + 2 \right) \qquad , \tag{E105}$$

and combining equations (E104) and (E105) turns into:

$$\sigma = \sigma_y \left(-2\frac{Y}{s}(c+x) + 1 \right) = \sigma_y \left(1 - 2Y \left(\frac{c+x}{s}\right) \right) \text{ for } -c < x < -b.$$
 (E106)



Figure E1.10: Stress and strain distribution curves of the second half-cycle.

Similarly, in the region a < x < c, the stress is equated by:

$$E\varepsilon = \sigma + \sigma_y \left(XY + \frac{2x}{s}Y - 1 \right)$$
(E107)

For x = c, the $\sigma = -\sigma_y$, so:

$$E\varepsilon = -\sigma_y + \sigma_y \left(XY + \frac{2c}{s}Y - 1 \right) = \sigma_y \left(XY + \frac{2c}{s}Y - 2 \right) \qquad , \tag{E108}$$

and combining equations (E107) and (E108) results into:

$$\sigma = \sigma_y \left(2\frac{Y}{s}(c-x) - 1 \right) = \sigma_y \left(2Y \left(\frac{c-x}{s}\right) - 1 \right) \quad \text{for } a < x < c \quad . \tag{E109}$$

Because the stress is constant in -b < x < a, an equation can be constructed for c by combining equation (E106) with x = -b and equation (E109) with x = a:

$$\left(1 - 2Y\left(\frac{c-b}{s}\right)\right) = \left(2Y\left(\frac{c-a}{s}\right) - 1\right) \to Y\left(\frac{2c-a-b}{s}\right) = 1 \qquad , \tag{E110}$$

and combining with equation (E95) for a and equation (E96) for b:

$$\frac{Y}{s}\left(2c-\frac{1}{2}s\left(\frac{1}{Y}-X\right)-\frac{1}{2}s\left(\frac{1}{Y}+X\right)\right) = \frac{Y}{s}\left(2c-s\left(\frac{1}{Y}\right)\right) = \frac{2cY}{s}-1 = 1 \rightarrow \frac{c}{s} = \frac{1}{Y}$$
(E111)

Equation (E111) can be combined with equation (E109) at x = a and equation (E95) for a to validate the observation at bullet 1, i.e. $E\varepsilon = \sigma_y XY$, for -b < x < a:

$$\sigma = \sigma_y \left(2Y \left(\frac{c-a}{s} \right) - 1 \right) = \sigma_y \left(2Y \left(\frac{1}{Y} - \frac{1}{2} \left(\frac{1}{Y} - X \right) \right) - 1 \right) = \sigma_y \left(2Y \left(\frac{1}{2Y} + \frac{X}{2Y} \right) - 1 \right) = \sigma_y \left(1 + XY - 1 \right) = \sigma_y XY \quad \text{for } -b < x < a \quad . \tag{E112}$$

The same result is obtained when equation (E106) with x = b, equation (E96) for b and equation (E111) are combined.

The plastic strains over the wall are distributed as follows. First, in the region -b < x < a, no plastic strain has developed during the first half-cycle or during the second half cycle:

$$\varepsilon_p^{1+2} = \varepsilon_p^1 + \varepsilon_p^2 = 0 + 0 = 0$$
 for $-b < x < a$. (E113)

In -c < x < -b, plastic strain has developed during the first half-cycle as given by equation (E99) and no plastic straining occurred during the second half-cycle (i.e. $\varepsilon_p^2 = 0$):

$$\varepsilon_p^{1+2} = \varepsilon_p^1 = \frac{\sigma_y}{E} \left(XY + \frac{2x}{s}Y + 1 \right) \qquad \text{for} \qquad -c < x < -b \quad . \tag{E114}$$

For x < -c, plastic strains developed during both first and second half-cycle. During the first half-cycle the plastic strain develops as described by equation (E99) due to compressive yielding. Part of this plastic strain is returned during the second half-cycle due to tensile yielding. This return of plastic strain during the second half-cycle progresses also as described by equation (E99), though it stops at x = -c, because yielding stops there. For that reason, the total plastic strain becomes:

$$\varepsilon_p^{1+2} = \varepsilon_p^1 + \varepsilon_p^2 = \frac{\sigma_y}{E} \left(XY - \frac{2c}{s}Y + 1 \right) \quad \text{for} \qquad x < -c \qquad . \tag{E115}$$

Combining equation (E115) with equation (E111) leads to:

$$\varepsilon_p^{1+2} = \frac{\sigma_y}{E} \left(XY - 2Y \frac{1}{Y} + 1 \right) = \frac{\sigma_y}{E} (XY - 1) \quad \text{for } x < -c \quad , \tag{E116}$$

which is indicated in figure E1.10 by "3".

The plastic strain development during the first and second half-cycles for the region a < x < c is similar to that of -c < x < -b, i.e. plastic straining occurs during the first half-cycle and not during the second half-cycle. Using equation (E102) results into:

$$\varepsilon_p^{1+2} = \varepsilon_p^1 = \frac{\sigma_y}{E} \left(XY + \frac{2x}{s}Y - 1 \right) \qquad \text{for} \qquad a < x < c \qquad . \tag{E117}$$

Subsequently, for x > c the plastic strain after the second half-cycle is:

$$\varepsilon_p^{1+2} = \varepsilon_p^1 + \varepsilon_p^2 = \frac{\sigma_y}{E} \left(XY + \frac{2c}{s}Y - 1 \right) \quad \text{for} \qquad x < -c \qquad . \tag{E118}$$

Combining equation (E118) with equation (E111) leads to:

$$\varepsilon_p^{1+2} = \frac{\sigma_y}{E} \left(XY + 2Y \frac{1}{Y} - 1 \right) = \frac{\sigma_y}{E} (XY + 1) \quad \text{for } x > c \quad , \quad (E119)$$

which is indicated in figure E1.10 by "4".

E6.3 Subsequent half-cycles

All subsequent half-cycles act the same as the first half-cycle in case thermal stress is applied or as the second half-cycle when the thermal stress is removed. Plastic strain does not accumulate (i.e. ratchetting does not occur), but the wall between $\frac{-s}{2} \le x < -c$ and between $c < x \le \frac{s}{2}$ is subject to cyclic plastic strains. This plastic cycling process is known as plastic shakedown (i.e low-cycle fatigue). The largest cyclic plastic strains ($\Delta \varepsilon_{p,max}$) occur at the surfaces of the wall. At the left wall (i.e. $x = \frac{-s}{2}$), the cyclic plastic strain is the combination of equations (E100) and (E116):

$$\Delta \varepsilon_{p,max}\left(x=\frac{-s}{2}\right) = \frac{\sigma_y}{E}(XY-Y+1) - \frac{\sigma_y}{E}(XY-1) = \frac{\sigma_y}{E}(2-Y).$$
(E120)

At the right wall (i.e. $x = \frac{s}{2}$), the cyclic plastic strain is equations (E103) and (E119) combined:

$$\Delta \varepsilon_{p,max}\left(x=\frac{s}{2}\right) = \frac{\sigma_y}{E}(XY+Y-1) - \frac{\sigma_y}{E}(XY+1) = \frac{\sigma_y}{E}(Y-2).$$
(E121)

E6.4 Boundary in Bree diagram between S₂ and P

Plastic shakedown does not occur when the cyclic plastic strain is zero. Taking equation (E120) leads to:

$$\Delta \varepsilon_p = \frac{\sigma_y}{E} (2 - Y) = 0 \qquad , \tag{E122}$$

which can only be achieved when:

Using equation (E53) to substitute Y results in:

$$\frac{\sigma_t}{\sigma_y} = 2 \to \sigma_t = 2\sigma_y \tag{E124}$$

Equation (E124) defines the boundary between elastic shakedown (S_2) and plastic shakedown (P) as shown in figure E1.1.

E7. Boundary in Bree diagram between Zone 1 and Zone 2

In the Bree diagram as shown in figure E1.1, all equations have been explained, except the one between zone 1 (i.e. S₁ and R₁) and zone 2 (i.e. S₂ and R₂). Zone 1 is the zone in which one surface yields during the first half-cycle. The ratchetting process in this zone, R₁, is described at ratchetting situation number 1 earlier in this chapter. The elastic shakedown in this zone, S₁, is the situation where initial yielding (i.e. during the first half-cycle only) occurs at one surface only, which is described earlier in the paragraph "Elastic shakedown". In zone 2, both surfaces are yielding during the first half-cycle. The ratchetting situation number 2 as explained before in this chapter. Elastic shakedown in this zone, S₂, is achieved when one surface yields initially (i.e. only during the first half-cycle) in tension and the other surface yields initially in compression.

The equation for the boundary between these two zones is obtained by using equation (E12), which is repeated here:

$$\sigma_1 = 2\sqrt{\frac{(\sigma_y - \sigma_p)}{\sigma_t}}\sigma_t - \sigma_y \qquad (E125)$$

As is best illustrated at the ratchetting situations, one surface yields when $\sigma_1 \leq \sigma_y$ and both surfaces yield when $\sigma_1 > \sigma_y$. This means that the boundary is situated at $\sigma_1 = \sigma_y$, which gives the following equilibrium in equation (E125):

$$\sigma_{y} = \sqrt{\frac{(\sigma_{y} - \sigma_{p})}{\sigma_{t}}} \sigma_{t} = \sqrt{\sigma_{t}(\sigma_{y} - \sigma_{p})} \to \sigma_{t}(\sigma_{y} - \sigma_{p}) = \sigma_{y}^{2} \qquad .$$
(E126)

E8. Other Bree diagrams

The Bree diagram of figure E1.1 is for a constant primary stress and a cyclic thermal stress. Other Bree diagrams are available as well and described briefly in this paragraph.

E8.1 Cyclic primary stress and constant thermal stress

This stress combination makes the occurrence of plastic shakedown or ratchetting not possible because there are no alternating plastic strains. The cycling is caused by the primary stress and in case plastic strain has developed during the first cycle, additional plastic strain does not develop during any of the subsequent cycles (i.e. the strain follows the same strain path from cycle to cycle). This results in elastic cycling or elastic shakedown situations only. For this situation, the Bree diagram is constructed as shown in the left figure of figure E1.11. In this diagram, S₁ is the elastic shakedown region in which the initial yielding occurs at one surface only. Elastic shakedown region S₂ is the region where both surfaces yield initially, one surface in tension and the other in compression.


Figure E1.11: Bree diagram for cyclic primary stress and constant thermal stress (left) and for cyclic primary stress and cyclic thermal stress in-phase (right) [E3].

E8.2 Cyclic primary stress and cyclic thermal stress (in-phase)

In the right figure of figure E1.11 is the Bree diagram shown of the combination of primary stress and thermal stress that cycle in-phase. Again, zone 1 is for the situation that one surface yields (initially) and zone 2 is for the cases that both surfaces yield (initially). An important observation is that with an increase in thermal stress this sequence of straining modes is always found: elastic cycling, elastic shakedown, plastic shakedown and finally ratchetting. Ratchetting occurs when the yielded zone of the loading half-cycle overlaps the yielded zone of the unloading half-cycle. Plastic shakedown is obtained when the surface(s) are subject to cyclic plastic strains, though the core of the wall remains elastic. Elastic shakedown is found when plastic strain develops at the surface(s) during the first half-cycle and no additional plastic strain develops during any of the subsequent cycles.

E8.3 Cyclic primary stress and cyclic thermal stress (out-of-phase)

For the greater part, the Bree diagram of the out-of-phase cyclic primary stress and cyclic thermal stress, as shown in figure E1.12, is quite similar to that of the in-phase Bree diagram of right figure of figure E1.11. The first main difference is that primary stress and thermal stress are not present at the same time and for that reason, the thermal stress can cycle elastically until yield strength regardless the amount of primary stress. Another main difference is that elastic shakedown only occurs when both surfaces yield initially, i.e. there is no S_1 region present. The last main difference is that the boundary between zone 1 and zone 2 is determined via different equation.



Figure E1.12: Bree diagram for cyclic primary stress and cyclic thermal stress out-of-phase [E3].

E9. Literature

- [E1] J. Bree, Elastic-plastic behaviour of thin tubes subjected to internal pressure and intermittent high-heat fluxes with application to fast-nuclear-reactor fuel elements, Journal of Strain Analysis, V2 N3, 1967.
- [E2] rickbradford.co.uk.
- [E3] S. Majumdar, Ratcheting problems for ITER, ANL/FPP/TM-253, Argonne National Laboratory, January 1991.

Appendix F – ASME Design Code

In this appendix, the background of the requirements and limits are explained in more detail.

The ASME II, the material standards code, has issued additional rules and regulations. These rules and regulations have resulted in a maximum allowable stress for design (see figure F1.1), which depends on creep strength and high temperature rupture strength, besides yield strength and ultimate tensile strength. The maximum allowable stress for design as defined by the ASME code, the creep strength and the rupture strength are dealt with later in this appendix.



Figure F1.1: ASME II allowable stress for Alloy 617 [F1].

Symbols

Besides the symbols as use in the main report, these symbols are used as well:

Coefficients for the model for fatigue according to ASME
Young's modulus at in the $k^{\rm th}$ cycle evaluated at mean temperature
Effective strain range in k th cycle
Elastic strain range in k th cycle
Equivalent plastic strain range in k th cycle
Multiplication factor on the average rupture stress at 100,000 hours
Plastic strain range component
Effective alternating equivalent stress (amplitude) for k th cycle
Range of peak equivalent stress of k th cycle

F1. Maximum allowable stress according to ASME design code

The values for allowable stresses in tables 1A and 1B in the ASME Section II, Part D, are mandatory for the design of ASME pressure retaining parts. These data are especially based on experience and results from service or from tests where the operating conditions are simulated with sufficient certainty. For new materials, test information is compared with available data obtained from successful applications

of similar materials. The ASME requires that the maximum allowable stress (σ_{allow}) is the lowest value obtained from the criteria of Table 1-100 (see table F1) [F2]. The descriptions and the equations for these stress values are described in the main report.

, Room Temperature and Below			Above Room Temperature							
Product/Material	Tensile Strength	Yield Strength		Tensile Strength		Yield Strength	Sti Rup	ress oture	Creep Rate	
Nrought or cast ferrous and nonferrous	<u>S7</u> 3.5	² / ₃ S _Y	$\frac{S_T}{3.5}$	$\frac{1.1}{3.5} S_T R_T$	2/3 SY	² / ₃ S _Y R _Y or 0.9S _Y R _Y ENote (1)]	F _{avg} S _{R avg}	0.8 <i>S_{R min}</i>	1.0 <i>S</i> c	
Welded pipe or tube, ferrous and nonferrous	$\frac{0.85}{3.5} S_T$	$\frac{2}{3} \times 0.85 S_{\gamma}$	$\frac{0.85}{3.5} S_T$	$\frac{(1.1 \times 0.85)}{3.5} S_T R_T$	2 ₃ × 0.85 S _Y	$\frac{2}{3} \times 0.85 S_Y R_Y$ or 0.9 × 0.85 $S_Y R_Y$ [Note (1)]	$(F_{\text{avg}} \times 0.85) S_{R \text{ avg}}$	$(0.8 \times 0.85)S_{R \min}$	0.85 <i>Sc</i>	

Table F1: Table 1-100 from mandatory appendix 1 of the ASME II [F2]

NOTE:

(1) Two sets of allowable stress values may be provided in Table 1A for austenitic materials and in Table 1B for specific nonferrous alloys. The lower values are not specifically identified by a footnote. These lower values do not exceed two-thirds of the minimum yield strength at temperature. The higher alternative allowable stresses are identified by a footnote. These higher stresses may exceed two-thirds but do not exceed 90% of the minimum yield strength at temperature. The higher values should be used only where slightly higher deformation is not in itself objectionable. These higher stresses are not recommended for the design of flanges or for other strain sensitive applications.

Nomenclature:

multiplication factor for the average creep rupture strength after 100,000 hours. $F_{ava} =$ - at T \leq 815 °C, F_{avg} = 0.67.

- at T > 815 °C, F_{avg} is determined from the slope of the log time to rupture versus log stress plot at 100,000 hours such that log $F_{avg} = 1/n_R$, but not larger than 0.67, with n_R = number equal to minus Δ log time to rupture divided by Δ log stress at 100,000 hours.

- $R_T =$ temperature-dependent trend curve value that determines the ratio between the temperature-dependent tensile strength to the room temperature tensile strength.
- temperature-dependent trend curve value that determines the ratio between the $R_Y =$ temperature-dependent of the yield strength to the room temperature yield strength.
- $S_C =$ average stress to produce a creep rate of 0.01%/1,000h.

 $S_{R,avg}$ = average creep rupture strength for rupture after 100,000 hours.

 $S_{R,min}$ = minimum creep rupture strength for rupture after 100,000 hours.

 $S_T =$ specified minimum tensile strength at room temperature [in MPa].

 $S_{Y} =$ specified minimum yield strength at room temperature [in MPa].

For components containing a longitudinal weld an extra safety margin of 15% applies, resulting in the multiplication factor of 0.85 in the equations for these components. Longitudinal welded components are not used in the boiler manufacturing and for that reason this safety margin is not regarded.

F1.1 Below the temperature of creep significance

The maximum allowable stress ($\sigma_{allow.}$) value for temperatures below the temperature from which creep and stress rupture strength govern the stress selection, is the lowest of the following:

- 1. ³/₃ of the specified minimum yield strength at room temperature.
- 2. ³/₃ or 90% of the yield strength at temperature, depending on the material.
- 3. The specified minimum tensile strength at room temperature divided by 3.5.
- 4. The tensile strength at temperature multiplied by 1.1 and divided by 3.5.

Some austenitic/non-ferrous materials at elevated temperatures (such as Alloy 617), a larger portion of the yield strength may be used for the allowable stress calculations because of the high proportionality limit, the high ductility and proven performance and safe use up to this higher stress value (i.e. the scatter in the data is low making the behaviour of these materials more predicable). The allowable stress may not exceed 90% of the minimum yield strength at temperature. The condition for this larger portion is that a small amount of plastic deformation is on its own not objectionable. This deviation of the allowable stress is not recommended to be used for the design of flanges or other strain sensitive applications as indicated by the footnote of Table 1-100 of ASME II (see also Table F1).

F1.1.1 Allowable stress from yield strength

These allowable stresses have been described and discussed in the main report paragraph 2.8 already. The maximum allowable stress based on yield strength at room temperature may not exceed $\frac{2}{3} \times 240$ MPa = 160 MPa for Alloy 617. The maximum allowable stress based on temperature-dependent yield strength is 0.9x the yield strength at the temperature considered.

F1.1.2 Allowable stress from ultimate tensile strength

The ultimate tensile strength for Alloy 617 at room temperature (for point no. 3) is minimal 655 MPa based on the stress tables from the ASME II. The maximum allowable stress based on ultimate tensile strength at room temperature may not be higher than 655 MPa/3.5 = 187 MPa. Since this value is independent of the temperature and already higher than the maximum allowable stress of $\frac{2}{3}$ of the yield strength at room temperature, this value may be disregarded for Alloy 617.The maximum allowable stress resulting from temperature-dependent ultimate tensile strength is 1.1x the result from the trendline value for temperature-dependent ultimate tensile strength x 655 MPa/3.5. The calculated values for this criterion are all higher than the maximum allowable stress values calculated for the temperature-dependent yield strength, so this criterion does not need to be considered any further.

F1.2 Above the temperature of creep significance

At temperatures above the temperature from which creep governs the stress selection, the maximum allowable stress is established by the minimum value of:

- 1. 100% of the average stress to produce a creep rate of 0.01%/1,000h.
- 2. 100 x F_{avg} % of the average creep rupture strength for rupture after 100,000 h.
- 3. 80% of the minimum creep rupture strength for rupture after 100,000 h.

High temperature stress values are based on representative uniaxial properties of the materials obtained under standard ASTM testing conditions or equivalent. No consideration is given for abnormal temperature or stress conditions, or for corrosive environments.

F1.2.1 Allowable stress from creep rate

According to ASME Section II, Part D, Appendix 1, the maximum allowable stress based on creep rate is the stress at which a creep rate ($\dot{\epsilon}$) of 0.01%/1,000h is exceeded. This is equivalent to 10⁻⁵ %/h or 10⁻⁷/h. Equation (2.3D) is rewritten to creep stress (σ_c) as a function of creep rate and temperature:

$$\sigma_{c} = \mu \left(\frac{\varepsilon_{allow}}{B_{c}} \exp\left(\frac{Q_{C}}{R \cdot T_{K}}\right)\right)^{1/n} , \qquad (F1A)$$

in which μ , n and B_c are temperature-dependent according to the relations described in paragraph 2.3. Q_c is a material constant (i.e. 410 kJ/mol) and R is the gas constant. The maximum allowable stress based on creep is equal to σ_c with $\dot{\varepsilon}_{allow}$ of 0.01%/1,000h.

F1.2.2 Allowable stress from creep rupture strength

ASME Section II, Part D, Appendix 1, states that the maximum allowable stress based on average creep rupture strength is a portion of the stress at which rupture occurs after 100,000 hour of exposure at a certain temperature. The portion is determined by a multiplication factor (F_{avg}), which is a safety factor. At T \leq 815 °C, the F_{avg} = 0.67, and for T > 815 °C, the F_{avg} depends on the reciprocal of the slope of the log time to rupture versus log stress plot at 100,000 hours, though it may never be larger than 0.67. Also, the Larson-Miller Parameter (*LMP*) has to be determined at a time to rupture (t_R) of 100,000 hours, and with the log t_R = log (100,000) = 5, equation 2.4B becomes:

$$LMP = 25T_{K}$$

(F1B)

with T_K in Kelvin. With the aid of this *LMP*, the rupture stress is now solely varying with temperature.

The values for F_{avg} above 815 °C are determined with the aid of the data plots on the right-hand side of figure F2.1. First, from this figure for each of the given temperatures the values of time to rupture and corresponding creep rupture strengths are determined. These values are displayed in table F2.



Figure F2.1: Stress versus 1% creep strain and time to rupture curves [F3].

The values from table F2 are used to construct graphs of the time to rupture as a function of the creep rupture strength. The trendlines of these graphs have the form:

$$t_R = a_R \sigma_{R,avg}^{n_R} , \tag{F1C}$$

in which t_R is the time to rupture in hours, a_R is a temperature-dependent material constant (relating time and stress, the unit is hour \cdot MPa^{-n_R}), $\sigma_{R,avg}$ is the average creep rupture strength in MPa, and n_R is a dimensionless temperature-dependent material exponent (i.e. the slopes of the graphs).

Time to rupt.	Rupture st	ress [MPa] a	t °C		
[h]	800 °C	850 °C	900 °C	950 °C	1000 °C
100	141	98	71	49	32
500	109	76	51	35	22.5
1,000	100	69	46	30	19
5,000	77	51	33	21	12.4
10,000	69	46	29	17.9	10.4
50,000	52	32.8	20	11.9	6.6
100,000	46	28.5	17	9.9	5.4

Table F2: Time to rupture versus applied stress for rupture at several temperatures.

The values of the material constants a_R and n_R are given in table F3:

Table F3: Values for a_R and n_R at several temperatures.

	800 °C	850 °C	900 °C	950 °C	1000 °C
a_R [hour · MPa ^{-n_R}]	2.050E+15	1.361E+13	1.074E+11	2.236E+09	7.241E+07
<i>n</i> _{<i>R</i>} [-]	-6.176	-5.546	-4.856	-4.312	-3.837

With these values equation F1C gives a good description of the relation between the t_R and the $\sigma_{R,avg}$, though it is evident that the slope is less steep (i.e. of the log t_R vs. log $\sigma_{R,avg}$ graph, which has the axis switched when comparing to figure F2.1) at a t_R of 100,000 hours. To obtain a more accurate slope value at 100,000 hours (= $n_R(10^5h)$), the values of both t_R and the $\sigma_{R,avg}$ are taken at 50,000 hours and 100,000 hours resulting in:

$$n_R(10^5h) = \frac{\log\left(\frac{t_{R;100,000h}}{t_{R;50,000h}}\right)}{\log\left(\frac{\sigma_{R,av;100,000h}}{\sigma_{R,av;50,000h}}\right)} ,$$
(F1D)

At temperatures above 815 °C, the ASME II, Part D, Appendix 1, states that F_{avg} is equated by:

$$\log F_{avg} = 1/n_R (10^5 h) \qquad , \qquad (F1E)$$

which is rewritten as:

$$F_{avg} = 10^{1/n_R(10^5h)}$$
 , (F1F)

The values for $n_R(10^5h)$ and F_{avg} at several temperatures are shown in table F4.

	, 0				
	800 °C	850 °C	900 °C	950 °C	1000 °C
$n_R(10^5h)$	-5.654	-4.933	-4.265	-3.767	-3.454
F_{avg}	0.665	0.627	0.583	0.543	0.513

Table F4: $n_R(10^5h)$ and F_{avg} at several temperatures.

Based on these data points, the slope at 100,000 hours $n_R(10^5h)$ can be calculated as a function of the temperature:

$$n_R(10^5h) = -8.800 \cdot 10^{-8}T^3 + 2.094 \cdot 10^{-4}T^2 - 1.513 \cdot 10^{-1}T + 2.637 \cdot 10^1$$
(F1G)

in which T is the temperature in degrees Celsius in the range 800 °C - 1000 °C. This equation is plotted in the following figure.



Figure F2.2: Relation between slope of log t_R – log $\sigma_{R,avg}$ graph at 100,000 h vs. temperature in °C.

Combining equations F1F and F1G gives the multiplication factor (F_{avg}) dependency on the temperature in the temperature range 800 °C - 1000 °C:

 $F_{avg} = 1.067 \cdot 10^{-8} T^3 - 2.823 \cdot 10^{-5} T^2 + 2.402 \cdot 10^{-2} T - 5.950 \cdot 10^0$,(F1H)

Also here, the temperature is in degrees Celsius. The plot is illustrated in figure F2.3.



Figure F2.3: relation between F_{avg} vs. temperature in °C.

By using equation (2.4A), the maximum allowable stress based on average rupture stress ($S_{R,avg}$) is:

$$S_{R,avg} = F_{avg} \cdot \sigma_{R,avg} = F_{avg} \cdot 4.53 \cdot 10^5 \exp(-3.41 \cdot 10^{-4} LMP)$$
, (F1J)

which after combining with equation (F1B) results in:

$$S_{R,avg} = F_{avg} \cdot 4.53 \cdot 10^5 \exp(-8.525 \cdot 10^{-3}T_K)$$
 , (F1K)

F1.2.3 Maximum allowable stress based on minimum creep rupture strength

According to ASME Section II, Part D, Appendix 1, the maximum allowable stress based on minimum creep rupture strength ($S_{R,min}$) at which rupture may occur after 100,000 hour of exposure at a certain temperature is 0.8 $S_{R,min}$. In turn, the minimum creep rupture strength is situated at 80% of the average creep rupture strength ($S_{R,avg}$) as is shown in the main report, paragraph 2.4.

Taking equation (2.4A), which is used to determine $\sigma_{R,avg}$ from the Larson-Miller Parameter, and equation (F1B) (i.e. fixed time to rupture of 100,000 hours, so $LMP = 25T_K$, with T_K in Kelvin) the value of the minimum creep rupture strength $\sigma_{R,min}$ is equated as:

$$\sigma_{R,min} = 0.8 \cdot \sigma_{R,avg} = 362512 \exp(-8.525 \cdot 10^{-3} T_K) , \qquad (F1L)$$

The allowable stress based on the $\sigma_{R,min}$ is referred to as $S_{R,min}$, is equal to 0.8 $\sigma_{R,min}$:

$$S_{R,min} = 0.64 \cdot \sigma_{R,avg} = 290010 \exp(-8.525 \cdot 10^{-3}T_K)$$
 , (F1M)

F2. Fatigue assessment in the ASME code

Standardised fatigue curves for Alloy 617 do not exist or are not acknowledged by the ASME yet. On the other hand, the ASME does allow the use of data from other sources, though conservative factors have to be included. The ASME has constructed a criteria document for 'Design by Analysis' in which the requirements for the construction of an ASME Code fatigue design curve are illustrated. This criteria document has been included into ASME, e.g. in ASME III, appendix I.

The construction of a fatigue design curve is based on strain-controlled tests. The data derived from these tests are plotted in a fatigue strain versus life (i.e. number of cycles to failure) graph and a best-fit curve is made, usually in the form of a Langer equation for the total strain amplitude (ε_a):

$$\varepsilon_a = C_A (N_f)^{-n_f} + C_B \qquad , \qquad (F2A)$$

in which N_f is the number of cycles to failure, and C_A , C_B and n_f are coefficients of the model.

A task group established by the ASME has submitted a proposal for the fatigue design curve of Alloy 617 for all temperatures below 425 °C [F5]. In the near future also a proposal for the Alloy 617 fatigue design curve for temperatures above 425 °C shall be issued.

F2.1 Loading histogram and loading categories

Loads vary with time during fatigue, thus a loading histogram must be developed to show the time variation of each specific load including all significant operating parameters, such as temperatures, pressures, supplemental loads, and the corresponding cycles/time. The number of cycles (e.g. start-ups/shutdowns, normal operation, upset conditions), the anticipated sequence of operation, and the relation/interaction between the applied loadings, have to be considered as well.

In the ASME, three kinds of loading categories are distinguished. Each loading category gives a set of loading conditions which is considered into the loading histogram as a separate event that each occur a number of times.

Each category is indicated by a level:

- Level A: Service loads. These are loads that evolve from normal operation of the system, so start-up, service, hot standby, and shutdown. These loads can commonly be described quite accurately.
- Level B: Anticipated deviation loads. Loads in this category are load transients that evolve from expected/anticipated single operator errors and/or control malfunction. Such load transients can be calculated into the expected life estimation.
- Level C: Incident loads. Incidental shutdown for correction of unstable and unacceptable loads and/or for repairs. This kind of shutdown is unexpected and has a low probability of occurrence and shall not result in gross loss of structural integrity of the component. The ASME has limited the number of these shutdowns to maximal 25 during lifetime.

F2.2 Cycle count methods

The API 579-1/ASME FFS-1 (edition 2007), is the standard used by API and ASME to gather the data and formulate the requirement to develop a design that is 'fit for service' (FFS). In Annex B3 this standard provides two cycle counting methods [F6], though an alternative cycle counting method may be used. These cycle-counting methods are also proposed in the ASME VIII, Div. 2, Annex 5B [F8]. In the main report is already described that max-min cycle-counting method considers non-proportional loads where the rain-flow method does not. Therfore, the max-min cycle-counting method is used.

F2.3 Screening criteria for fatigue

Fatigue analysis is not required in the ASME when the provisions of one of three options are met. The first option is based on experience with comparable equipment. The sponsor does not have sufficient experience on the use of austenitic material, especially on nickel-based material and this option is disregarded for that reason. Method A is the second option and can solely be used for materials with a specified minimum ultimate tensile strength at room temperature of \leq 552 MPa. Nickel-based alloy 617 has a specified minimum ultimate tensile strength at room temperature of 655 MPa, and therefore method A is not valid. Method B remains as option and can be used for all materials. All criteria of Method B have to be met to supersede a fatigue analysis and these criteria are described in appendix C, paragraph C4.5.1. In case any of the screening criteria is not fulfilled, a detailed fatigue analysis is required. The aim is to always consider fatigue in the design analysis and for that reason this screening is not performed nor incorporated into the Toolbox.

F2.4 Fatigue assessment

In case a fatigue analysis must be performed, an elastic-plastic (E-P) analysis ensures that plastic strain effects are incorporated. This process is already described in appendix C, appendix C4.5.2.

F2.5 ASME restrictions on the accumulated creep-fatigue

The design code does not deal with the interaction between creep and fatigue qualitatively, though it recognises the significance of this interaction. This is described already in paragraph 2.8 of the main report.

F3. Literature

- [F1] J.K. Wright, L.J. Carroll, C.J. Cabet, Ti. Lillo, J.K. Benz, J.A. Simpson, W.R. Lloyd, J.A. Chapman, R.N. Wright, Characterization of Elevated Temperature Properties of heat Exchanger and Steam Generator Alloys, Nuclear Engineering and Design, Vol. 251, p. 252-261, 2011.
- [F2] ASME Boiler & Pressure Vessel Code, Section II Part D: Properties (metric) Materials, ASME, 2011.
- [F3] H. Nickel, F. Schubert, H.J. Penkalla, H.H. Over, Mechanical Design Methods for High Temperature Reactor Components, Nuclear Engineering and Design 76, p. 197-206, Amsterdam (NL), 1983.
- [F4] Special Metals, Inconel Alloy 617, 2005. http://www.specialmetals.com/documents/Inconel%20alloy%20617.pdf
- [F5] R. Wright, VHTR Materials Overview, DOE-NE Materials Crosscut Coordination Meeting, Idaho National Laboratory, July 2013.
- [F6] API 579-1/ASME FFS-1, "Fitness-For-Service", American Society of Mechanical Engineers and American Petroleum Institute, June, 2007.
- [F7] ASME III, Rules for construction of Nuclear Facility Components, American Society of Mechanical Engineers, 2013.
- [F8] ASME VIII, Div. 2, *Rules for construction of Pressure Vessels, Alternative Rules,* ASME Boiler and Pressure Vessel Code, 2013.

Appendix G – Method for testing creep-fatigue interaction

The interaction between damage mechanisms creep and fatigue may be investigated further. This can be done by determining the mechanical properties pertaining to creep-fatigue damage. This timedependent damage may be determined from constant-amplitude strain-controlled tests in which fatigue testing is performed with slow strain rates or with cycles involving longer hold-times (i.e. creep has time to develop).

G1. Creep-fatigue test data

From creep-fatigue tests, data that can be determined are:

- Cyclic stress-strain response (i.e. hysteresis diagram);
- Relaxation deformation response (i.e. measuring declining stress);
- Cyclic creep deformation response (note: this test requires stress-control);
- Cycles to the formation of crack(s), i.e. life-time determination;
- Cyclic hardening or softening response;
- Determination of activation energy for creep-fatigue mechanism based on the loss of energy in the hysteresis (this as function of the frequency).

G1.1 Cyclic stress-strain deformation response

The stress-strain hysteresis diagram for a continuous cycling fatigue test (i.e. without hold-time), which included some plastic deformation is determined as is shown as an example in figure G1.



Figure G1: Stress-strain hysteresis diagram without any hold-time [G1].

In the stress-strain hysteresis diagram, these symbols are used:

- *E_c* Compressive elastic modulus;
- *E_T* Tensile elastic modulus;
- $\Delta \varepsilon_{in}$ Inelastic strain range (sum of plastic strain and creep strains);
- $\Delta \varepsilon_p$ Plastic strain range;
- $\Delta \varepsilon_t$ Total strain range;
- σ_{max} Maximum stress in the cycle;
- σ_{min} Minimum stress in the cycle;
- $\Delta \sigma$ Stress range (2x the stress amplitude).

G1.2 Relaxation deformation response

With the introduction of a hold-time in the tension direction of a strain-controlled test, the resulting stress reduces due to stress relaxation, which is caused by the transfer of elastic strain into inelastic creep strain. The stress-strain hysteresis diagram changes in a manner as is shown in the left figure of figure G2. The hold-time is chosen in tension direction because more damage is expected to accumulate over time under exposure of tensile stresses than under exposure of compressive stresses (i.e. worst case scenario is tested). This should come forward in a faster relaxation rate for tensile loads than for compression loads.



Figure G2: Stress-strain hysteresis diagram of a strain-controlled test with a hold-time in tension (left) and stressstrain hysteresis diagram of a stress-controlled test with a hold-time in tension (right) [G1].

G1.3 Cyclic creep deformation response

The cyclic creep deformation response is determined in a stress-controlled test with a hold-time in tension direction. Stress is applied up to σ_{max} and held for some time. During this hold-time at constant stress, inelastic creep strain accumulates until the stress is released. This process is shown in the right figure of figure G2. The reason for choosing a hold-time in tensile direction instead of compression direction is the same as for the relaxation deformation response (see previous paragraph).

G1.4 Cycles to the formation of cracks

The number of cycles to crack formation is the number of cycles to the attainment of a specific crack size or cracked area. This can be determined in a strain-controlled creep-fatigue test as a specific percentage decrease in the maximum cyclic tensile stress (σ_{max}) relative to the stationary maximum cyclic stress level during the test. The tensile stress is measured because cracks initiate and progress due to tensile stresses while compression stresses close the cracks and partly heal the plastically deformed material in front of the crack tip. This stress decrease is shown as X% in the left figure of figure G3.



Figure G3: Maximum cyclic stress vs. number of cycles diagram showing the formation of cracks (left) and an example of maximum stress vs. number of cycles diagram of a material showing continuous cyclic softening (right) [G1].

In figure G3, N_f is the number of cycles to failure which represents the formation of a crack.

G1.5 Cyclic hardening or softening response

The cyclic hardening or softening is determined as the respective increase or reduction of the tensile stress (σ_{max}) or stress range ($\Delta\sigma$) with the number of cycles in a strain-controlled creep-fatigue test. An example for a material with continuous cyclic softening is shown in the right figure of figure G3.

G2. Cycle shape

For creep-fatigue testing a number of cycle shapes may be considered, though the cycle shapes that reflect the service conditions for boiler components most accurately are those with a hold-time of the strain in tension direction (see left figure of figure G4) and the cycle with hold-time of the strain (ε) in both tension and compression direction (see right figure of figure G4).



Figure G4: Cycle with hold-time in tension direction in a strain-controlled test (left) and cycle with hold-time in both tension and compression direction in a strain-controlled test (right) [G1].

G3. Relations and correlations in the stress-strain hysteresis diagram

An empirical relation exists for the correlation of isothermal creep-fatigue data. This relation is called the frequency-modified strain-life. Commonly this relation is taken at half-life, since this cycle gives the most proper averaged result. In case the test data exhibit measurable inelastic strains and there is no significant change in the time-dependent mechanism in the set of data being correlated, good correlations can be found based on this relation [G1]. The frequency-modified strain-life modifies the number of cycles to failure (N_f) with the frequency (ν) of the cycle:

$$\Delta \varepsilon_{in} = C_{cf} \left(N_f \cdot v_{cf}^{k_{cf}-1} \right)^{-\beta} , \qquad (G1A)$$

in which $\Delta \varepsilon_{in}$ is the inelastic strain range and C_{cf} , β and k_{cf} are constants that are determined from a linear regression analysis of the test data plot in a $\log(\Delta \varepsilon_{in})$ versus $\log(N_f)$ graph. This plot reduces to a time-independent strain-life relation in case $k_{cf} = 1$.

The wave-form can affect the damage mechanism and hence influence the frequency-modified strainlife. The actual frequency is the inverse of the total time for one cycle, though different segments of the cycle may influence the type of damage mechanism and/or the amount of damage differently. For example, compressive holds in ductile materials can partially heal creep damage developed during the tension hold. This leads to a diminished influence of the frequency on life.

G4. Apparatus

In order to perform the tests described above, the following apparatus and equipment is required.

G4.1 Test machine

A servo-controlled tension-compression fatigue machine has to be used that has:

- Sufficient lateral stiffness to maintain accurate alignment and to avoid bending.
- A smooth start-up.
- No backlash when passing through zero force.

G4.2 Force transducer

The force transducer and adherent electronics have to be compatible with the test machine (e.g. suitable for tension-compression testing). Therewith, it has to fulfill the following requirements:

- The axial and lateral rigidity has to be high to ensure an accurate alignment and to avoid bending.
- Axial forces have to be measured with accuracies smaller than 1% of the reading [G1].
- Temperature compensated with zero drift or sensitivity variation of maximum 0.002% per °C of the full scale [G1].

G4.3 Loading train and specimen fixtures

The test specimen fixtures have to be aligned in such a manner, that the major axis of the test specimen coincides closely with the force axis throughout each cycle in order to minimise bending strains. The accuracy of the alignment has to be kept consistent from specimen to specimen. It is advisable to keep bending strains within 5% of the minimum axial strain range. Two forms of misalignment of the loading train that may result in bending of the specimen are shown in figure G4.



Figure G4: Two types of loading train misalignment that may lead to specimen bending [G1].

The loading bars have to be equipped with grips to fix the test specimen. The design of these grips has to be such, that motion is not lost at the test specimen/grip interface when the stress displacement is going through zero force. This is achieved by these grip characteristics:

- A surface ensuring the specimen axis alignment.
- A loading surface through which the load is transmitted in one direction.
- A second loading surface through which the load is transmitted in reverse direction.
- Continuous contact of the specimen with both loading surfaces.

The choice of grip configuration for the mounting of the specimen depends on the test specimen configuration and this is dealt with in paragraph G5 about the test specimen.

The fixtures must be cooled in order to limit heat transfer from the hot zone to the force transducer. Means of cooling include water cooling of the coils or jackets and/or forced air cooling of installed fins at the outer ends of the loading bars. Care has to be taken to make sure that force transducer calibration and load train alignment are not affected by the presence of the cooling devices.

G4.4 Extensometer

The extensometer has to meet these requirements:

- Suitable for measuring dynamic displacements over long periods, i.e. it shall have a rapid response with a low hysteresis (commonly smaller than 0.1% [G1] of the extensometer output).
- Minimal slippage, drift and instrument hysteresis.

The extensometer has to measure longitudinal extension and is therefore installed parallel to the test specimen. In order to minimise the forces between the test specimen surface and the extensometer probe tips, the extensometer has to be supported independently.

G4.5 Heating system

Heating of the specimen using a resistance furnace is the best option because the specimen can be heated uniformly. Other options are less suitable due to larger heat exposure of the extensometer (for radiant furnace heating), magnetic radiation that may interfere with the other equipment (for induction heating) or high costs and complexity of test set-up (for inert gas or liquid heating).

The heating system has to have a set-up that ensures the uniform heating of the test specimen to the required temperature. This is obtained when the temperature gradient across the gauge section is maximum the greater of 2 °C or 1 percent of the nominal test temperature (in °C) throughout the duration of the test [G1]. The controlled temperature has to be maintained within ± 2 °C during the complete test [G1]. Undesirable gradients and fluctuations in the temperature due to draughts must be avoided.

G4.6 Temperature measurement

Thermocouples are placed in contact with the surface of the test specimen and used to monitor the specimen's temperature.

G4.7 Cycle counter

The number of cycles should be recorded by the data acquisition system with a resolution of 1% or better than the actual cycle-life [G1], although usually the cycle count is done flawlessly.

G4.8 Data recording

A recording system, preferably automatic and digital, has to be used that is capable of collecting and simultaneously processing of the force, displacement and temperature data as function of time and cycles. The frequency of the force-displacement-time sampling has to be sufficient to guarantee a good definition of the hysteresis loop and hold-time transients. Important is to obtain the values of the force and extension at the hysteresis loop turning points (e.g. at start and end of hold-time, and at cycle maxima and minima).

G5. Test specimen

Creep-fatigue tests are performed with round bar test specimens subject to uniaxial loading. The uniform gauge section of the test specimen is commonly used for creep-fatigue testing using an extensometer installed parallel to the specimen axis. The geometry of this test specimen is shown in figure G5.



Figure G5: Uniform gauge test specimen for creep-fatigue testing [G1].

In figure G5, the symbols and sizes used are:

$d \ge 5mm$	diameter of cylindrical gauge section.
$l_o \ge 1.5d$	gauge length.
$r \ge 2d$	transition radius (from gauge section to grip-end).
$D_G \geq 2d$	diameter of grip-ends.

Also, the parallel portion of the parallel gauge section test specimen (L_0) has to be longer than the extensioneter gauge length (l_0) , though the difference between these two lengths has to be smaller than d to limit the chance of failure outside the extension gauge length.

The tolerances on parallelism, concentricity and perpendicularity have to remain lower than 0.01 mm relative to the axis or reference plane [G1].

G5.1 End connection of test specimen

The dimensions of the end connections are dependent on the testing machine that is being used, though often the button-head connection is used at the end of the test specimen with a diameter of 3d. This configuration is shown in the left figure of figure G6.



Figure G6: Button-head end connection of test specimen (left) and fixture for button-head test specimen (right) [G1].

The fixture of the test specimen in which a good alignment is obtained and back lash is avoided is shown in the right figure of figure G6.

G5.2 Preparation of test specimen

The test specimens have to be prepared as shown in figures G5 and G6. Additionally, the following preparations and actions have to be done to make the test specimen ready for testing:

- Identifying by engraving a marking on the end connections at both sides of the specimen.
- Final machining and polishing of the specimen until the mean roughness is ≤ 0.2 µm and all circumferential scratches are removed.
- Dimensional check of the test specimen.
- Attachment of the thermocouples by spot welding.

G6. Test procedure

Prior to the start of the test, the test specimen has to be mounted into the test machine and some preliminary measurements have to be done. First, the test specimen has to be mounted into the loading grips taking into account the alignment of the assembly, avoiding damage to the surface of the test specimen and avoiding handling that may affect the properties of the material (e.g. prestraining of the specimen). Then the extensometer is attached along the test specimen's gauge length. The cycle shape, maximum and minimum strain values, strain rate, hold-times and temperature are entered as input parameters into the software program of the testing machine.

G6.1 Measurements prior to start of test

Preliminary measurements have to be performed in order to identify possible problems with force or displacement or within the measuring systems. Prior to the start of the actual test, these checks have to be performed:

- Elastic modulus determination at room temperature and after heating to the test temperature. These values shall not deviate by more than 10% of the expected values to ensure that the test results are representative.
- Coefficient of thermal expansion after the temperature is stabilised at the test temperature, which is determined by dividing the extension measured by the extensiometer by the temperature change while the test machine does not apply any force. Also for the coefficient of thermal expansion, the measured value shall not deviate by more than 10% of the expected value to make sure that the test results are representative.
- The gauge length of the extensometer has to be compensated/corrected to account for the gauge length extension due to thermal expansion. This is done by resetting the output signal of the extensometer (i.e. put to zero).

G6.2 Heating of the specimen

The test specimen is heated to the test temperature and maintained at the temperature for a minimum of 30 minutes prior start of the test. During heating the tolerances in temperature as described in paragraph G4.5 shall not be exceeded. The (resulting) stress inside the test specimen is not allowed to be higher than 10% of the material's yield strength at test temperature [G1].

G6.3 Testing

The test is performed as follows:

- Output of the extensometer has to be put to 'zero' without any force on the test specimen.
- The direction of the first part of the cycle is in tension because possible hold-times are in tension direction as well.
- The strain rate applied during the test is the standard rate of 1.10⁻³ s⁻¹ [G1].
- The load cycle is applied via the cycle wave-form desired (see paragraph G2), including possible hold-time(s). Commonly, the test is strain-controlled.
- The stability of the strain measurement has to remain within 0.5% from cycle to cycle and within 2% during the entire test [G1].
- The controlled temperature has to be maintained within ±2 °C of the desired temperature in the complete test specimen and during the entire test [G1].

G6.4 Recording of test results

During testing, these parameters need to be recorded continuously:

- Force, extension, temperature and perhaps crack condition at the critical turning points (i.e. at maximum and minimum force and extension, and at start and end of hold-times) within the cycle.
- Initial force-extension hysteresis loop at start of the test.
- Periodic the hysteresis loops in logarithmic increments of the number of cycles (i.e. 1, 2, 5, 10, 20, 50, etc.) [G1].

G6.5 End of the test

The end of the test is defined as a 10% decrease in the maximum tensile stress in relation to the maximum tensile stress level as determined during the test progression [G1]. As an example, this decrease in maximum tensile stress is shown in figures G4 and G5, where it is indicated with the letter "X". This criterion is entered in order to avoid damage to both extensometer and fracture surface which may be caused by complete separation of the two test specimen halves.

In case much cyclic hardening or softening occurs, another end-of-test criterion may be more suitable. Since hardening and softening occurs commonly in both tension and compression direction, the ratio between these two maximum stresses may be used. The test end is achieved when the maximum tensile stress has a certain percentage decrease in relation to the maximum compression stress.

G6.6 Shutdown after completion of test

The furnace has to be switched off directly after completion of the test. This is done to limit the extent of oxidation of the test specimen's outer surface and crack surfaces, which makes post-test examination more difficult. Care has to be taken to avoid overloading of the test specimen during cooling.

G7. Post-test examination

First, a macroscopic examination has to be performed to determine the location(s) of the cracks.

Via metallographic examination (i.e. optical microscopy and scanning electron microscopy on both fracture surface and cross-sectional cut samples), the creep and fatigue damage fractions need to be determined. An example of a characteristic of creep damage is voids in the material. Fatigue damage can for example be observed by striations on the fracture surface.

G8. Records

The test records should contain the following information:

- 1. Material specification, i.e. Alloy 617 and reference to the material heat used.
- 2. Test specimen identity (i.e. unique sample number).
- 3. Cycle shape, including the strain rate.
- 4. Hold-time characteristics (like position, duration and control parameter).
- 5. Test temperature.

- 6. Modulus of elasticity and coefficient of thermal expansion.
- 7. Characteristics of first cycle:
 - a. Maximum and minimum strain.
 - b. Resulting maximum and minimum stress.
 - c. Strains at start and end of hold-time.
 - d. Resulting stresses at start and end of hold-time.
 - e. Inelastic strain range.
- 8. Characteristics of half-life cycle (same as for the first cycle) with additional:
 - a. Cycle number at half-life.
- 9. Characteristics of cycle when the crack formation criterion is reached, which is same as for the first cycle, though with additional:
 - a. Number of cycles to crack formation.
 - b. Description of crack formation criterion.
- 10. Details of post-test examination:
 - a. Fracture locations.
 - b. Damage mechanisms observed and their fractions.
- 11. Any reasoning about the validity of the performed test, e.g. description of possible events that occurred during testing that may affect the test results.

G9. Considerations

When performing the creep-fatigue tests and evaluating the test results, some considerations have to be taken into account:

- It is advisable to perform continuous cycling fatigue tests and creep tests as well at the same temperatures and using the same material (i.e. from the same heat number, so with equal chemical composition, grain sizes, heat treatments, etc.). With this data, the influences of creep and fatigue on the creep-fatigue test results can be distinguished more accurately.
- 2. Oxidation can also be responsible for important interaction effects relating to damage accumulation, especially at elevated temperatures, unless such tests are performed in an inert environment or in vacuum.

G10. Literature

[G1] Standard Test Method for Creep-Fatigue Testing, ASTM E2714-13, ASTM International, 2013.