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Experimental characterisation of low-cycle fatigue behaviour of thin-walled and solid WAAM-fabricated high-strength steel under variable mean strain

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ABSTRACT

Wire Arc Additive Manufacturing enables efficient fabrication of large-scale steel components with high geometric freedom. However, reliable structural use of WAAM high-strength steels requires an in-depth understanding of their cyclic behaviour, particularly in the low-cycle fatigue regime, which is dominated by plastic deformation. This study investigates the LCF behaviour of WAAM-fabricated ER100S-G high-strength steel using specimens extracted from thin-walled (WAAM-W) and solid deposits (WAAM-S). Strain-controlled fatigue tests were conducted at room temperature under strain ranges $\Delta\epsilon$, of 0.35% to 2.00%. WAAM-W was tested at strain ratios $R_\epsilon = -1, 0$ and 0.5 while specimens from the WAAM-S configuration were evaluated at $R_\epsilon = 0$.

Although WAAM-S exhibited higher monotonic and cyclic strength, it underwent more pronounced cyclic softening, with a reduction of approximately 35% in the 0.2% offset cyclic yield stress, compared with 17% for WAAM-W. WAAM-S showed a steeper elastic strain-life slope, indicating that its strength advantage diminished with increasing fatigue life, while WAAM-W exhibited greater fatigue ductility and plastic-strain accommodation. This strength stability trade-off is attributed to the heterogeneous and banded bainitic morphology of WAAM-S, which provides cyclic strain localisation. In contrast, the more homogeneous WAAM-W microstructure and strain-induced transformation of retained austenite contribute to distributed plasticity and improved cyclic stability. For WAAM-W, positive mean strain reduced fatigue life at lower strain amplitudes due to a persistent tensile mean stress, whereas increased plasticity at higher amplitudes contributed to stress relaxation. The SWT formulation captured the mean-strain dependence of fatigue life, while the energy-based analysis linked fatigue life to the elastic and plastic strain-energy contributions.

1. Introduction

The demand for durable, high-performance structural components supporting material-efficient designs has grown rapidly in sectors such as civil engineering, offshore structures, energy infrastructure, and maritime. Wire Arc Additive Manufacturing (WAAM) has emerged as a promising solution capable of meeting these requirements [1]. Owing to its relatively high deposition rate compared with other metal additive manufacturing processes and its highly controllable process parameters, WAAM enables the efficient fabrication of large-scale metallic components with complex, free-form geometries, facilitating structural optimisation [2,3]. This flexibility also provides precise control over the deposited material, allowing the production of components with varying

thicknesses and locally tailored material properties through parameter adjustments during deposition [4].

Despite these advantages, the effective utilisation of WAAM ultimately depends on achieving consistent and reliable mechanical performance across a wide range of manufacturing conditions. A critical aspect of this reliability is the behaviour of the material under cyclic loading, as structural components are subjected to repeated environmental and operational loads throughout their service life. Consequently, fatigue strength becomes a primary design criterion for WAAM structural components.

The factors governing fatigue performance are closely linked to the choice of process parameters and deposition strategy, as these determine: (i) the thermal history, as heat input and cooling rate control the

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resulting microstructure including grain size, grain orientation, texture, dislocation density, and phase transformations [5–7]; (ii) the formation of process-inherent imperfections such as porosity, lack of penetration, and as-built surface roughness [8–10], which act as preferential crack-initiation sites; (iii) the development of residual stresses arising from inhomogeneous thermal cycles during deposition [11], which may act as mean stresses superimposed on the cyclic load; and (iv) the resulting mechanical properties of the deposited material [12]. The combined influence of these factors can locally induce cyclic plasticity even under relatively low nominal stresses. This effect becomes most severe at critical locations where geometric stress concentrations coincide with process-induced imperfections, thereby accelerating damage accumulation and ultimately governing premature fatigue failure.

High-strength steels (HSS) are widely used in structural applications due to their favourable strength-to-weight ratio, which enables reductions in component size and structural self-weight while maintaining high load-carrying capacity. These characteristics lead to significant material savings and improved structural efficiency, particularly in large-scale infrastructure applications [13]. When implemented in WAAM-fabricated components, these advantages may be further amplified. Extensive research has focused on the fundamental aspects of WAAM-fabricated high-strength steels, including process calibration [14–16], microstructural characterisation [17–19], residual stress evolution [20], and monotonic mechanical properties [20–23]. However, the reliable application of WAAM-fabricated HSS in structural components requires a clear understanding of their behaviour under cyclic loading conditions. Given these considerations, accurate characterisation of cyclic elastic–plastic material properties through assessment within the low-cycle fatigue (LCF) regime is essential for enabling realistic mechanical analysis and reliable fatigue predictions for WAAM components.

Several studies have examined the influence of process-related factors, including surface condition, build orientation, and microstructural heterogeneity, on the fatigue behaviour of WAAM-fabricated materials. Most of these investigations have focused on titanium alloys, stainless steels, and carbon steels, particularly ER70S-6. For machined specimens extracted from thick-walled ER70S-6 deposits, the fatigue lives remained consistently below those of conventionally manufactured steels, with no significant influence of specimen orientation or through-thickness extraction location, underlining the role of fabrication route in the resulting LCF response [24]. Thin-walled ER70S-6 plates produced using a GMAW-based WAAM process also exhibited lower fatigue resistance than conventionally manufactured S355 steel in the machined condition, with a further reduction in the as-built condition attributed to orientation-dependent surface undulations associated with the deposition path [25]. Building on this work, Fang and Zong [26] compared the LCF behaviour of WAAM ER70S-6 and ER110S-G steels in both machined and as-built conditions. Under comparable surface conditions, ER70S-6 exhibited superior LCF resistance to ER110S-G, indicating that higher material strength does not necessarily translate into improved strain-controlled fatigue performance. Furthermore, machined WAAM 316L stainless steel specimens showed pronounced anisotropy in LCF behaviour, with specimens loaded perpendicular to the build direction exhibiting higher fatigue performance than those loaded parallel to it [27]. These studies highlight the need to consider fabrication route, deposition parameters, thermal history, build geometry, and specimen orientation when interpreting the LCF behaviour of WAAM materials.

While these studies provide valuable insight into the cyclic behaviour of commonly investigated WAAM alloys, the fatigue behaviour of WAAM high-strength steels remains unexplored. The existing studies have primarily focused on the high-cycle fatigue (HCF) regime [28–30]. This gap is particularly significant since the fatigue resistance of high-strength steels does not increase proportionally with their static strength [31], making them more sensitive to cyclic plastic deformation and defect-driven crack initiation under plasticity-dominated loading conditions.

To address these challenges, the present study experimentally investigates the low-cycle fatigue behaviour of WAAM-fabricated ER100S-G high-strength steel using specimens extracted from thin-walled and solid deposits. Strain-controlled fatigue tests are performed under multiple strain ratios to evaluate the influence of mean strain on cyclic deformation behaviour and fatigue life. In addition, microstructural characterisation using electron backscatter diffraction (EBSD) and fractographic analysis using scanning electron microscopy (SEM) are employed to reveal the mechanisms of fatigue crack initiation, propagation, and final fracture. Through combined cyclic constitutive characterisation, strain–life analysis, and energy-based evaluation, this study provides new insight into the fatigue behaviour of WAAM high-strength steels and contributes to the development of reliable fatigue assessment approaches for additively manufactured structural components.

In this paper, firstly, the material and adopted methods are described, including the WAAM processing conditions, specimen preparation, and strain-controlled fatigue testing. Then, the experimental results and corresponding discussion are presented, including the cyclic stress–strain response, cyclic softening behaviour, and microstructural characterisation based on EBSD analysis. Finally, the fatigue life prediction approaches are calibrated, including the application of strain–life and energy-based models.

2. Materials and Methods

2.1. Materials

The material used in this study is a low-alloy high-strength steel wire, commercially designated as 3Dprint AM 70 (ER100S-G), supplied by *voestalpine Böhler Welding* and specifically developed for WAAM applications [32]. The detailed chemical composition of the wire, together with that of the substrate plates, is listed in Table 1. A wire diameter of 1.2 mm was used in combination with an M21 shielding gas, consisting of 82% argon and 18% CO₂.

The material was deposited using a pulsed Gas Metal Arc Welding (GMAW-P) process to produce the two configurations, as shown in Fig. 1:

- Thin-walled configuration (WAAM-W): fabricated by vertically stacking a single weld bead (with an average bead width of 10 mm) layer-by-layer, representative of thin-walled structural geometries. The WAAM-W walls were printed on 10 mm-thick S690 QL substrate plates.
- Solid configuration (WAAM-S): produced by depositing four adjacent weld beads per layer with 35% overlap to form a filled bulk geometry with an average thickness of approximately 50 mm. The WAAM-S blocks were printed on 30 mm-thick S690 Q+A substrate plates.

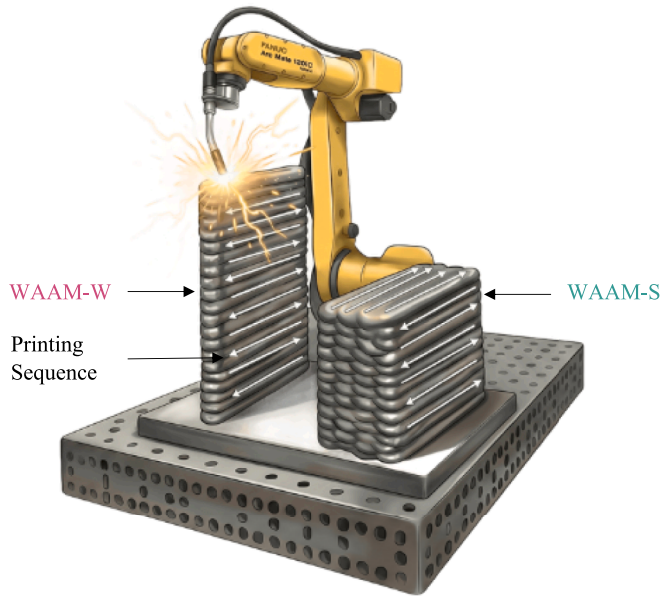
To promote a more uniform thermal distribution and minimize distortion during fabrication, an alternating deposition strategy was adopted in which the start and end points of successive layers were reversed for both configurations. The selected deposition strategy for the WAAM-S configuration was based on findings from a previous comprehensive study on solid specimens [10], in which different infill strategies and printing sequences with varying heat inputs were systematically evaluated. That study demonstrated that alternating deposition paths improved material homogeneity and reduced defect formation, resulting in a more consistent microstructural and mechanical response. The welding process parameters employed for printing both WAAM-W and WAAM-S configurations are summarized in Table 2. It should be noted that the interpass temperature was reduced from 250 °C to 150 °C during the fabrication of the WAAM-S configuration. This adjustment was implemented to better control heat accumulation, enhance process efficiency in terms of the overall deposition rate, and enhance the resulting mechanical performance.

Monotonic tensile tests were performed on specimens extracted from

Table 1

Chemical composition of Böhler 3Dprint wire AM 70 and substrates (%).

	C	Mn	Si	S	P	Cr	Mo	Ni	Fe
3Dprint wire									
ER100S-G	0.08	1.7	0.6	0	0	0.2	0.5	1.5	Bal.
Substrates									
S690 QL	0.08	1.65	0.39	0.0009	0.008	0.18	0.23	0.18	Bal.
S690 Q + A	0.17	1.5	0.44	0.0009	0.012	0.037	0.01	0.028	Bal.

**Fig. 1.** Schematic illustration of the Wire Arc Additive Manufacturing (WAAM) process used to fabricate WAAM-W and WAAM-S deposits.**Table 2**

Welding Process Parameters used for the fabrication of WAAM-W and WAAM-S configurations.

Type	WAAM-W	WAAM-S
Process	GMAW-P	
Current [A]	150	230
Voltage [V]	21.3	24.9
Travel Speed [mm/s]	5	5
Wire Feed speed [m/min]	4.7	7.7
Heat Input* [kJ/mm]	0.51	0.92
Interpass Temperature [°C]	250	150

* using a thermal efficiency factor of 0.8, in accordance with ISO 1011-1 for MIG/MAG welding.

WAAM-W and WAAM-S. The specimens were extracted in two orientations relative to the WAAM building direction: horizontal, parallel to the printing direction, and vertical, perpendicular to the printing direction,

Table 3

Measured mechanical properties of WAAM-W, WAAM-S, and ER100S-G wire.

	WAAM-W		WAAM-S		ER100S-G Wire
Orientation	Horizontal	Vertical	Horizontal	Vertical	
Yield strength [MPa]	565	566	748	744	720
Tensile strength [MPa]	735	731	803	811	780
Elongation [%]	23	25	17	14	16

to evaluate the influence of anisotropy. The average mechanical properties obtained for each orientation and configuration are presented in Table 3, together with the nominal properties of the ER100S-G feedstock wire. The results indicate that anisotropy in the mechanical properties is negligible, confirming that the selected deposition strategy ensured relatively uniform material properties throughout the deposited material. The WAAM-S specimens exhibit an average yield strength of 746 MPa, which is approximately 32% higher than the average yield strength of 565.5 MPa observed for the WAAM-W specimens. Compared with the ER100S-G wire, which has a yield strength of 720 MPa and tensile strength of 780 MPa, the WAAM-S configuration shows slightly higher strength, whereas the WAAM-W configuration exhibits lower strength but considerably higher ductility. The elongation of WAAM-W reaches 23–25%, compared with 16% for the wire and 14–17% for WAAM-S.

2.2. Low-cycle fatigue tests

2.2.1. Specimen preparation

The WAAM deposits, both WAAM-W and WAAM-S, were machined by CNC milling followed by precision grinding on both sides to remove the as-built surface irregularities and obtain flat plates with controlled geometry. Flat LCF specimens, designed in accordance with the ASTM E606 standard [33], were extracted in two orientations relative to the printing direction: vertical and horizontal. The final specimen geometry, along with the extraction layout, is presented in Fig. 2.

To minimise surface-induced effects on fatigue behaviour, the gauge sections of all specimens were polished. The polishing procedure consisted of sequential grinding using SiC papers (80–2000 grit), followed by final polishing with a 3 µm diamond suspension, resulting in a mirror-like surface finish. Surface roughness was measured using a Mitutoyo Surftest SJ-210 with a cutoff length λ_c of 2.5 mm, yielding an arithmetic average roughness $R_a = 0.03 \mu\text{m}$ and an average maximum height $R_z = 0.245 \mu\text{m}$.

2.2.2. Experimental procedure

Low-cycle fatigue (LCF) tests were conducted in accordance with ASTM E606 under strain-controlled loading at room temperature using a Schenck 600 kN servo-hydraulic testing machine. A sinusoidal waveform was applied, and the test frequency was adjusted for each strain amplitude to maintain a constant average strain rate of $0.8\% \text{ s}^{-1}$. The strain was measured and controlled using a calibrated Instron extensometer with a 12.5 mm gauge length. The primary test series was conducted at a strain ratio of $R_\epsilon = 0$, defined as $R_\epsilon = \epsilon_{\min}/\epsilon_{\max}$, for both WAAM-W and WAAM-S configurations. Specimens extracted in both horizontal (H) and vertical (V) orientations relative to the build direction were tested. A series of total strain ranges, $\Delta\epsilon$, ranging from 0.35% to 2.0%, was investigated. For the baseline ($R_\epsilon = 0$) series, multiple specimens were tested at selected strain levels for both WAAM-W and WAAM-S to evaluate repeatability and characterise the scatter in the strain–life response. Supplementary tests were conducted on the WAAM-W configuration under fully reversed loading ($R_\epsilon = -1$) and positive mean strain ($R_\epsilon = 0.5$) to evaluate the influence of mean strain. These tests were based on one specimen per condition; therefore, the mean-strain results should be interpreted as indicative trends, considering the limited number of repetitions and the associated experimental

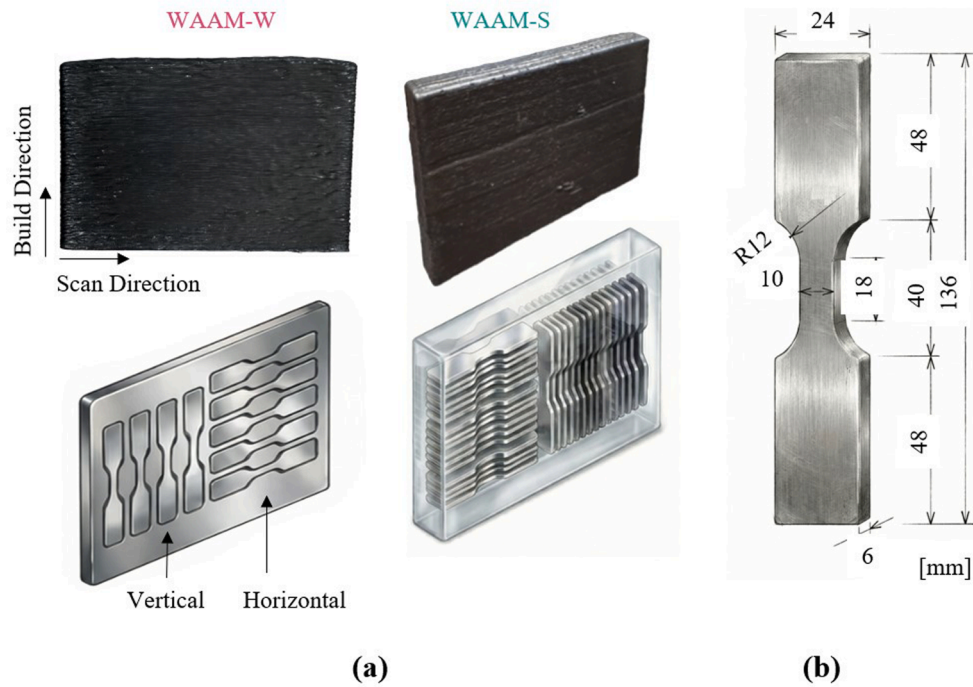


Fig. 2. Fabrication and testing of WAAM specimens. (a) As-built WAAM-W and WAAM-S configurations and extracted fatigue specimens. (b) Geometry and dimensions of the machined fatigue specimen according to ASTM E606 (all dimensions in mm).

scatter and uncertainty. The LCF test matrix is summarised in Table 4, including the WAAM configuration, strain ratio R_ϵ , specimen orientation, applied total strain range, and number of repetitions.

Throughout all tests, hysteresis loops were recorded at selected cycle intervals to monitor the evolution of the cyclic stress–strain response. The fatigue life, N_f , was defined as the cycle at which the stress amplitude, $(\Delta\sigma/2)$, decreased by 15% relative to the stabilised cyclic stress amplitude. This criterion was adopted to provide a conservative and consistent definition of failure across specimens with different degradation responses.

Fig. 3a shows the normalised stress-amplitude evolution for two representative specimens, denoted as Curve 1 and Curve 2, together with the corresponding hysteresis loops at selected reduction levels, in Figures Fig. 3b and Fig. 3c, respectively. The two curves exhibit different degradation behaviour, particularly in the descending part of the response: Curve 1 shows a relatively short degradation stage after stabilisation, whereas Curve 2 exhibits a more gradual decrease followed by a longer degradation stage. This indicates that both the onset and rate of cyclic degradation can vary between specimens. The corresponding hysteresis loops support the use of the 15% stress amplitude reduction criterion. At a 10% reduction, the loops remain close to the stabilised response, indicating that the mechanical response is still only moderately affected by damage development. At a 15% reduction, clear deviations from the stabilised loop become visible, marking the onset of a more pronounced degradation stage. This change is particularly evident in Fig. 3c, where the loop shape changes more markedly as the stress amplitude decreases. At a 25% reduction, the distortion becomes more pronounced, indicating more advanced damage development.

Table 4
LCF test matrix.

Configuration	R_ϵ	Orientation	$\Delta\epsilon$ [%]	Repetitions
WAAM-S	0	H,V	0.4–2.0	See Table 5
WAAM-W	0	H, V	0.35–2.0	See Table 6
	–1	H,V	0.6–2.0	1
	0.5	V	0.6–2.0	1

Therefore, the 15% reduction criterion was interpreted as a conservative crack-initiation-type criterion, defining fatigue life before the measured response becomes strongly affected by extensive crack propagation and severe hysteresis-loop distortion.

2.3. Microstructural Observation

Specimens with approximate dimensions of $10 \times 10 \times 6 \text{ mm}^3$ were extracted from the WAAM-W and WAAM-S deposits for microstructural characterisation. The samples were prepared using standard metallographic procedures [34], including grinding and polishing, to obtain a surface suitable for microscopic examination. Initial observations were performed using a Keyence VHX7000 digital optical microscope to assess the layer morphology and overall structural homogeneity of the WAAM deposits. Subsequently, detailed crystallographic characterisation was carried out using Electron Backscatter Diffraction (EBSD) on a JEOL IT800SHL scanning electron microscope equipped with an EBSD detector. EBSD scans were performed with a step size of $0.05 \mu\text{m}$ to evaluate crystallographic orientation, grain morphology, and phase distribution.

2.4. Fractography

Following the low-cycle fatigue (LCF) testing, fracture surfaces were examined to identify crack initiation sites and characterize the dominant failure mechanisms associated with the different specimen configurations and strain amplitudes. For SEM examination, selected specimens were sectioned from the gauge region and ultrasonically cleaned in trichloroethylene prior to analysis. Fractographic observations were performed on specimens from both WAAM-W and WAAM-S configurations, covering representative low and high strain amplitude conditions.

3. Results and Discussion

This section presents and discusses the experimental results of the strain-controlled fatigue tests together with the corresponding microstructural characterisation. The cyclic deformation behaviour, cyclic

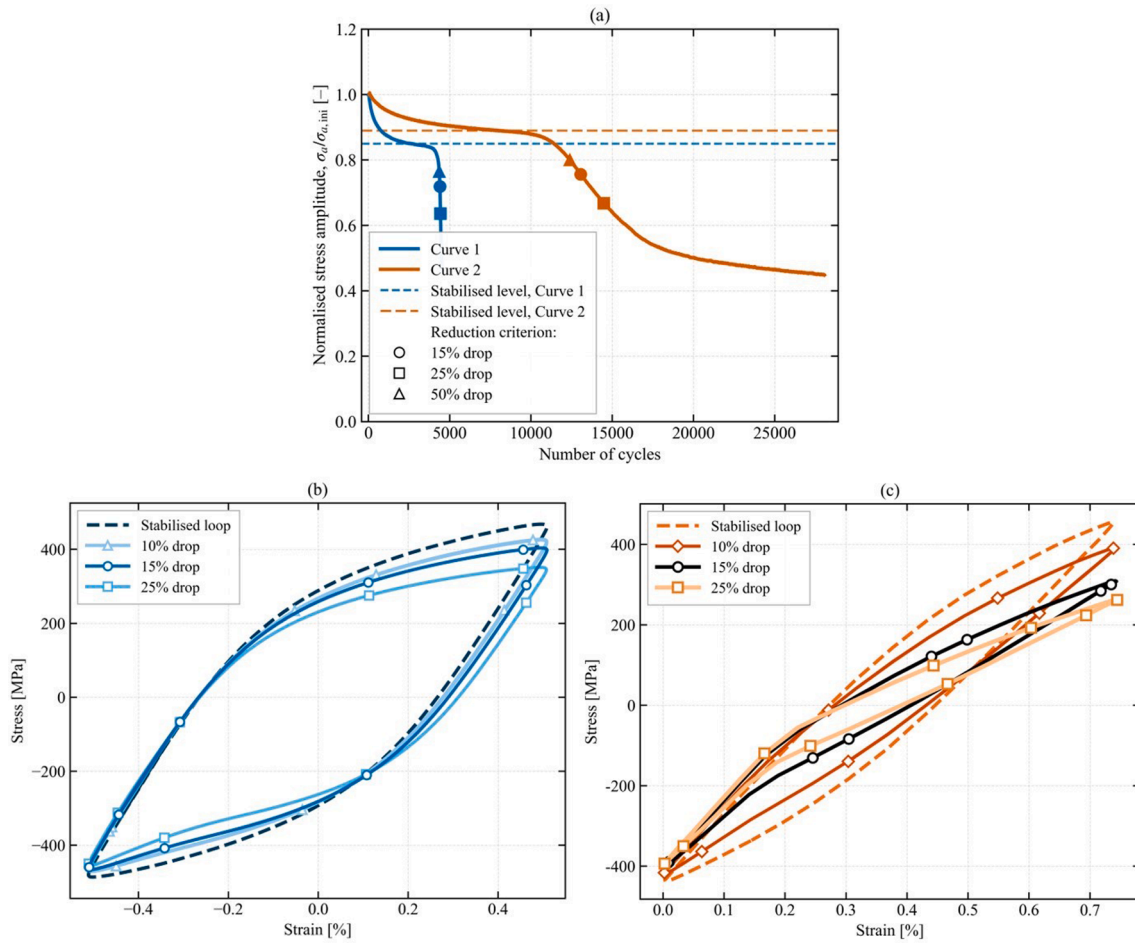


Fig. 3. Definition of the fatigue-life criterion based on the reduction in stress amplitude.

stress-strain curve, and microstructural features are analysed for the WAAM-W and WAAM-S configurations.

3.1. Experimental results

The results of the strain-controlled fatigue tests conducted at $R_e = 0$ for WAAM-S and WAAM-W configurations are summarized in Table 5 and Table 6, for each orientation. The tables include the applied total strain range $\Delta\epsilon$, the corresponding number of cycles to failure N_f , and

Table 5
Fatigue test results of WAAM-S ($R_e = 0$).

Specimen	$\Delta\sigma$ [MPa]	σ_m [MPa]	$\Delta\epsilon$ [%]	$\Delta\epsilon_p$ [%]	$\Delta\epsilon_e$ [%]	N_f [Cycles]
S-H-2%-1	1177.3	-7.7	2.02	1.37	0.65	801
S-H-2%-2	1174.4	-8.8	2.03	1.40	0.63	683
S-H-1.5%-1	1131.8	-9.2	1.52	0.91	0.61	1619
S-H-1%-1	1056.1	0.0	1.00	0.57	0.43	4375
S-H-1%-2	1010.8	-0.2	1.00	0.57	0.43	5676
S-H-0.8%-1	1000.6	6.8	0.80	0.40	0.40	7477
S-H-0.8%-2	987.1	11.4	0.80	0.37	0.42	7942
S-H-0.55%-1	966.0	46.6	0.55	0.26	0.29	15,241
S-H-0.4%-1	757.6	231.1	0.40	0.15	0.25	130,200
S-V-2%-1	1203.4	-11.7	2.01	1.35	0.66	674
S-V-1%-1	1047.2	0.9	1.00	0.56	0.44	3986
S-V-1%-2	1055.6	-1.3	1.00	0.58	0.42	3867
S-V-0.8%-1	1002.5	19.4	0.80	0.39	0.41	7143
S-V-0.8%-2	1031.8	8.6	0.80	0.39	0.41	6775
S-V-0.55%-1	922.2	60.2	0.53	0.26	0.28	20,216
S-V-0.4%-1	791.9	258.4	0.40	0.15	0.25	81,411
S-V-0.4%-2	746.0	275.0	0.40	0.15	0.25	56,090

Table 6
Fatigue test results of WAAM-W ($R_e = 0$).

Specimen	$\Delta\sigma$ [MPa]	σ_m [MPa]	$\Delta\epsilon$ [%]	$\Delta\epsilon_p$ [%]	$\Delta\epsilon_e$ [%]	N_f [Cycles]
W-H-2%-1	1183.5	-16.4	2.03	1.40	0.63	518
W-H-1.8%-1	1135.2	-12.7	1.82	1.20	0.62	1102
W-H-1.2%-1	1006.0	-4.0	1.20	0.76	0.44	3813
W-H-1%-1	945.2	-2.9	0.94	0.40	0.54	6475
W-H-1%-2	954.9	-5.1	0.94	0.43	0.51	5731
W-H-1.2%-2	1015.9	-6.1	1.22	0.67	0.56	1905
W-H-1%-1	947.4	-6.7	0.93	0.42	0.51	5340
W-H-1%-2	948.4	-4.9	0.94	0.42	0.52	2859
W-H-1%-3	966.1	-8.8	1.00	0.47	0.53	5424
W-H-0.75%-1	916.3	2.8	0.74	0.21	0.53	12,492
W-H-0.75%-2	894.5	12.5	0.74	0.15	0.59	20,835
W-H-0.75%-3	901.0	0.9	0.74	0.23	0.51	10,265
W-H-0.35%-1	622.1	243.8	0.35	0.00	0.35	367,239
W-H-0.34%-2	617.1	185.3	0.34	0.00	0.34	562,112
W-V-2%-1	1143.3	-10.5	2.03	1.42	0.61	692
W-V-1.8%-1	1131.0	-9.9	1.82	1.24	0.58	1243
W-V-1.2%-1	1002.4	-8.8	1.22	0.70	0.53	2993
W-V-1.2%-2	1027.5	-5.2	1.20	0.78	0.41	3144
W-V-1%-1	967.0	7.6	0.95	0.38	0.56	6258
W-V-0.75%-1	920.2	-2.8	0.74	0.26	0.48	7891
W-V-0.75%-2	902.6	9.5	0.73	0.18	0.56	13,714
W-V-0.61%-1	866.0	-2.8	0.61	0.16	0.45	9940
W-V-0.34%-1	608.5	249.5	0.34	0.00	0.34	505,685
W-V-0.34%-2	739.3	101.8	0.34	0.01	0.33	124,305

the stabilised stress-strain characteristics derived from the hysteresis loops at half-life ($0.5N_f$). The stress range ($\Delta\sigma$) was directly extracted from the stabilised hysteresis loops. The plastic strain range ($\Delta\epsilon_p$) was

defined as the horizontal width of the loop measured at the stabilised mean stress level. The elastic strain range ($\Delta\epsilon_e$) was subsequently obtained by decomposing the total strain range into elastic and plastic components.

Each specimen is identified using a designation that includes the configuration, build orientation, applied total strain range, and number of specimens. For example, W-H-1%-1 refers to a wall specimen (W) extracted in the horizontal orientation (H), i.e., parallel to the printing direction, tested under a 1% total strain range, and representing specimen number 1.

3.2. Cyclic Stress–Strain Response and Hysteresis Loop Evolution

The cyclic stress–strain response of the WAAM-W and WAAM-S configurations under a total strain range of $\Delta\epsilon = 0.8\%$ and $R_e = 0$ is presented in Fig. 4

a–b. Both configurations show cyclic softening, characterised by a progressive reduction in stress amplitude with increasing number of cycles. However, the magnitude of softening differs significantly between the two configurations. The WAAM-S material shows a reduction

of approximately 19% between the first and stabilised cycles, whereas the WAAM-W configuration exhibits a smaller decrease of about 13%, indicating a significantly lower cyclic stability in the WAAM-S configuration.

This difference is reflected in the stabilised hysteresis loops, where the WAAM-S configuration develops a substantially larger plastic strain range. At $\Delta\epsilon = 0.8\%$, the stabilised plastic strain range reaches approximately $\Delta\epsilon_p \approx 0.39\%$ for WAAM-S, while the WAAM-W configuration exhibits a considerably smaller value of $\Delta\epsilon_p \approx 0.21\%$. The larger cyclic plastic strain accumulation observed in the WAAM-S configuration confirms a greater degradation of cyclic resistance to plastic deformation. A similar trend is observed across the remaining strain ranges, where the WAAM-S configuration consistently develops a larger plastic strain range than the WAAM-W configuration.

The stabilised hysteresis loops for specimens extracted along the horizontal and vertical build orientations are shown in Fig. 4

c–d. The nearly overlapping loops indicate that the cyclic plastic response is essentially independent of the build direction for both WAAM-W and WAAM-S configurations. This suggests that the adopted deposition strategy effectively mitigates anisotropy in the material's

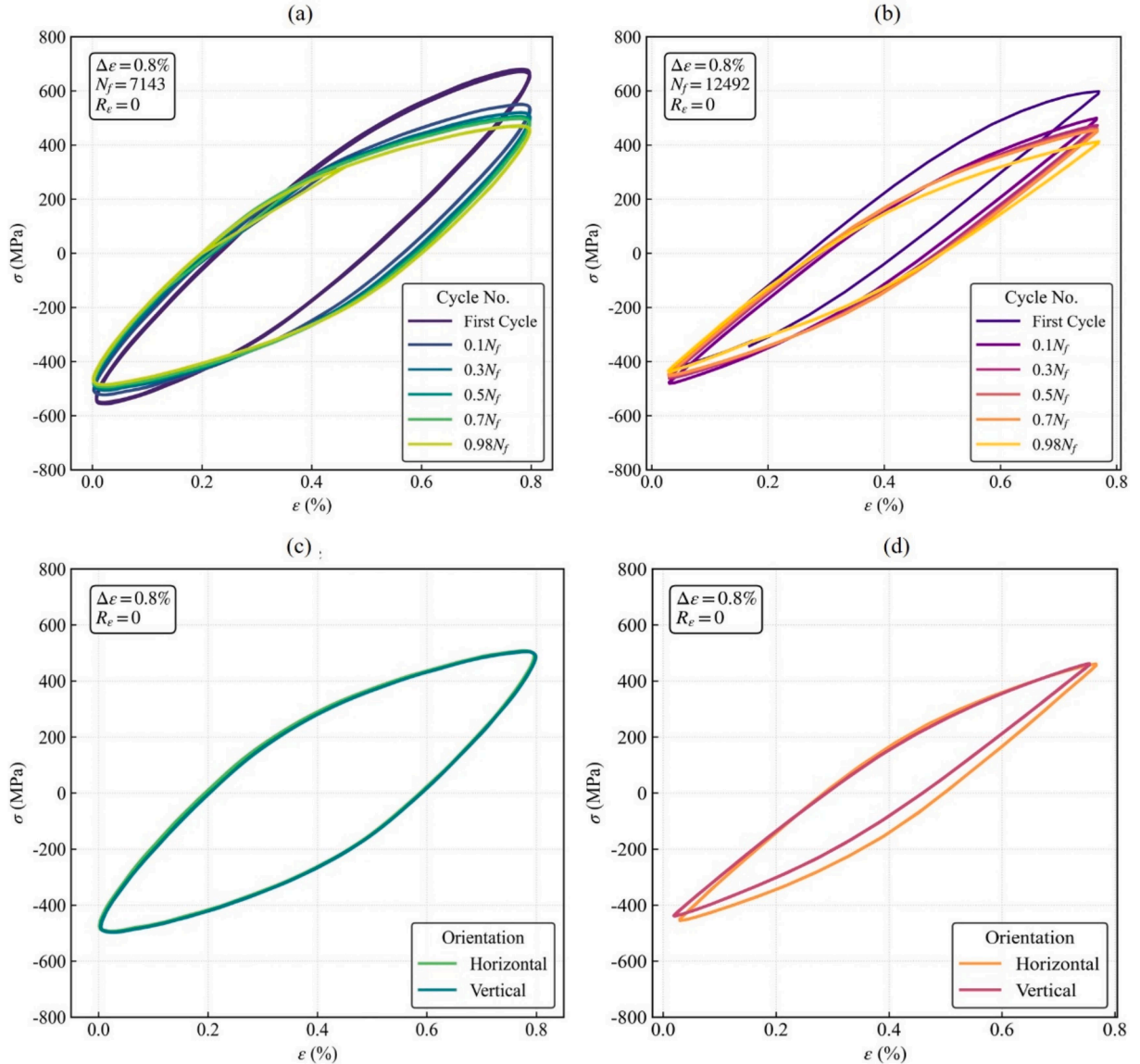


Fig. 4. (a–b) Evolution of the stress–strain hysteresis loops with increasing cycle number at $\Delta\epsilon = 0.8\%$ under $R_e = 0$ loading for (a) WAAM-S and (b) WAAM-W configurations. (c–d) Stabilised hysteresis loops at half-life (0.5 N_f) for the different orientations: horizontal and vertical of (c) WAAM-S and (d) WAAM-W specimens.

cyclic deformation behaviour.

The evolution of stress amplitude with increasing number of cycles for all investigated strain amplitudes is presented in Fig. 5a–b. The cyclic stress response follows a characteristic three-stage evolution consisting of an initial rapid softening stage, followed by a prolonged period of gradual stabilised softening, and a final sharp reduction in stress amplitude associated with macro-crack growth and stiffness degradation. The early-stage reduction in stress amplitude is consistent with the evolution of the dislocation substructure under cyclic loading, where rearrangement and partial annihilation of dislocations reduce the internal resistance to plastic flow. To quantify the softening behaviour, a cyclic softening ratio (SR) is defined based on Eq. (1) as the relative change in stress amplitude between the first cycle ($\sigma_a^{N=1}$) and the stabilised mid-life cycle ($\sigma_a^{N=0.5N}$):

$$SR = \frac{\sigma_a^{N=1} - \sigma_a^{N=0.5N}}{\sigma_a^{N=1}} \quad (1)$$

The variation of SR with total strain range is presented in Fig. 6. The results reveal a pronounced dependence of cyclic softening on the imposed strain amplitude. For both configurations, the softening ratio increases with increasing strain range and reaches a maximum around $\Delta\epsilon \approx 1.0\%$, after which a slight decrease is observed as the strain range approaches $\Delta\epsilon \approx 2.0\%$. This trend indicates that the relative reduction in stress amplitude does not increase proportionally with increasing plastic deformation at the highest strain amplitudes.

Across the entire strain range, the WAAM-S configuration consistently exhibits higher cyclic softening ratios than WAAM-W. The softening ratio for WAAM-W remains within approximately $SR \approx 0.10$ – 0.15 , whereas WAAM-S reaches values of $SR \approx 0.17$ – 0.22 , indicating a substantially greater reduction in stress amplitude between the first and stabilised cycles. This behaviour reflects enhanced cyclic plastic strain accumulation in the WAAM-S configuration, resulting in accelerated microstructural degradation mechanisms, including localised strain concentration and microstructural rearrangement. Despite its higher monotonic strength, the WAAM-S configuration therefore exhibits lower cyclic stability, indicating that monotonic strength is not a sufficient predictor of fatigue resistance under low-cycle loading conditions.

3.3. Stabilised Cyclic Stress–Strain Curve

The intrinsic cyclic constitutive behaviour of the WAAM-W and WAAM-S configurations was characterised using the Ramberg–Osgood relationship evaluated at half of the fatigue life:

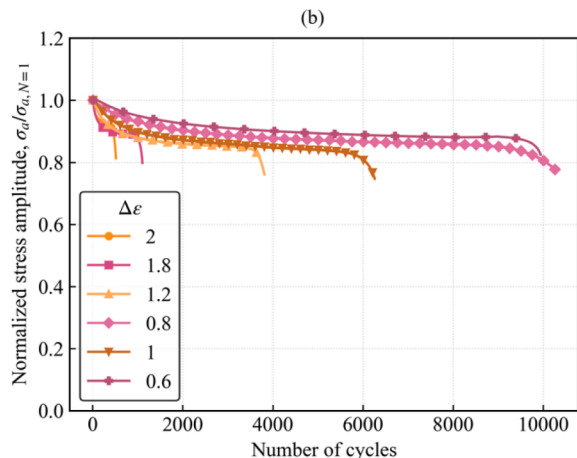
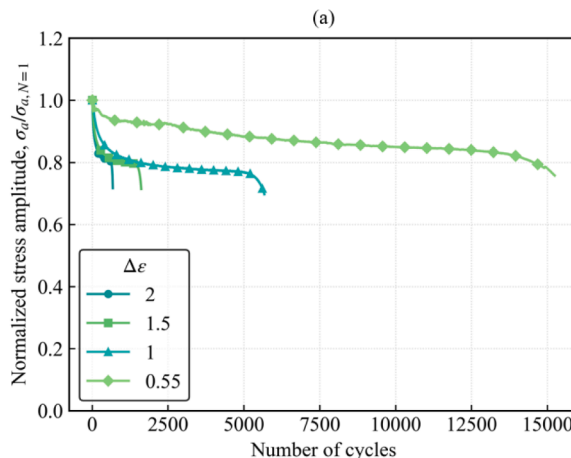


Fig. 5. Evolution of the normalized stress amplitude with the number of cycles for strain-controlled fatigue tests conducted on the WAAM-S (a) and WAAM-W (b) at $R_e = 0$.

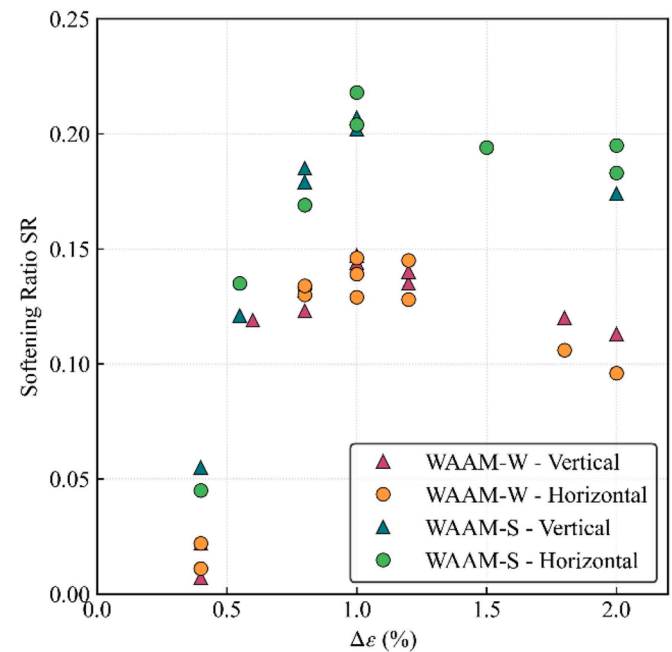


Fig. 6. Cyclic softening ratio as a function of total strain range $\Delta\epsilon$ for WAAM-W and WAAM-S specimens at $R_e = 0$

$$\frac{\Delta\epsilon}{2} = \frac{\Delta\epsilon_e}{2} + \frac{\Delta\epsilon_p}{2} \quad (2)$$

$$\epsilon_a = \frac{\sigma_a}{E} + \left(\frac{\sigma_a}{K'}\right)^{1/n'} \quad (3)$$

where σ_a and ϵ_a represent the stress and total strain amplitudes of the stabilised cycle, respectively, E is the elastic modulus, and K' and n' denote the cyclic strength coefficient and cyclic strain hardening exponent. The parameters K' and n' were determined from log–log regression of the plastic strain amplitude against the corresponding stress amplitude, as illustrated in Fig. 7. Because the strain-controlled tests were conducted at $R_e = 0$, non-zero mean stresses developed at the lowest strain amplitudes. These data points were therefore excluded from the fitting procedure to ensure that the extracted parameters represent the intrinsic cyclic plastic response of the material.

The resulting cyclic parameters reveal a clear distinction between the two WAAM configurations. The WAAM-S specimens exhibit consistently

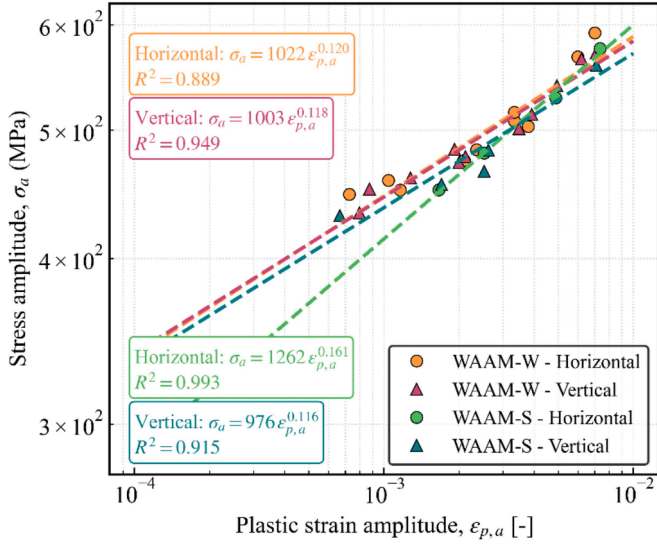


Fig. 7. Plastic strain amplitude versus stress amplitude in log-log scale for WAAM-W and WAAM-S for the different building orientations.

higher cyclic strength coefficients ($K' = 1109\text{--}1281$ MPa) compared with the WAAM-W configuration ($K' = 1003\text{--}1022$ MPa), indicating that higher stress amplitudes are developed for a given plastic strain amplitude. In addition, the larger strain hardening exponents obtained for the WAAM-S configuration ($n' = 0.128\text{--}0.151$) compared with WAAM-W ($n' = 0.118\text{--}0.120$) correspond to a steeper stress–plastic strain relationship.

With respect to building orientation, as shown in Fig. 7, WAAM-W specimens exhibit nearly identical cyclic stress–strain responses for the horizontal and vertical directions, indicating negligible orientation dependence of the stabilised cyclic behaviour. For the WAAM-S configuration, a moderate difference is observed, with the vertical specimens exhibiting slightly higher cyclic strength coefficients and strain-hardening exponents. Nevertheless, the overall influence of build direction on the cyclic constitutive parameters remains limited under the investigated loading conditions.

The cyclic stress–strain curves derived from the Ramberg–Osgood parameters are compared with the corresponding monotonic responses in Fig. 8. The two configurations exhibit distinctly different cyclic

behaviour. For WAAM-S, the cyclic curve remains below the monotonic response throughout the investigated strain range, indicating cyclic softening at all tested amplitudes. In contrast, the WAAM-W cyclic curve lies below the monotonic response at lower strain amplitudes but crosses above it beyond approximately 1% strain amplitude, indicating a transition from cyclic softening to cyclic hardening.

This strain-amplitude-dependent response of WAAM-W is consistent with observations reported for other WAAM carbon steels. Machined thick-walled WAAM ER70S-6 exhibited cyclic softening at strain amplitudes below approximately 0.6% and cyclic hardening at higher amplitudes, with limited dependence on specimen orientation and through-thickness extraction location [24]. A similar response was reported for machined WAAM ER70S-6 and high-strength ER110S-G, for which the cyclic stress–strain curves exceeded the corresponding monotonic responses beyond approximately 1% strain [26]. WAAM-W therefore follows the amplitude-dependent softening-to-hardening transition previously observed in WAAM carbon steels. In contrast, WAAM-S shows no such transition and remains cyclically softened throughout the investigated range. By comparison, conventional rolled S690 steel, with a nominal strength level similar to the AM70 feedstock, has been reported to exhibit a comparatively stable cyclic response over a substantial portion of fatigue life [31]. This contrast further indicates that the pronounced cyclic degradation of WAAM-S is related not only to strength level but also to the thermal history and microstructural state produced by the multi-bead deposition strategy.

The distinction between the two WAAM configurations is further reflected in the reduction of the 0.2% offset cyclic yield stress, which was approximately 35% for WAAM-S and 17% for WAAM-W. This indicates that WAAM-S retains higher cyclic strength but exhibits lower cyclic stability, consistent with a material state more susceptible to continued degradation under cyclic plastic deformation.

3.4. Influence of Strain Ratio on the Cyclic Response of WAAM-W

The influence of mean strain ratio on the fatigue response was assessed using the WAAM-W configuration under $R_\epsilon = -1, 0$, and 0.5 . As shown in Fig. 9a, the cyclic softening ratio exhibits no systematic dependence on the applied mean strain ratio. This observation is consistent with the stabilised cyclic stress–strain behaviour discussed earlier and indicates that the imposed mean strain does not significantly alter the intrinsic cyclic constitutive response of the material. The role of the mean strain ratio is instead reflected in the evolution of mean stress during cycling. At a strain range of 0.6%, which lies within the more

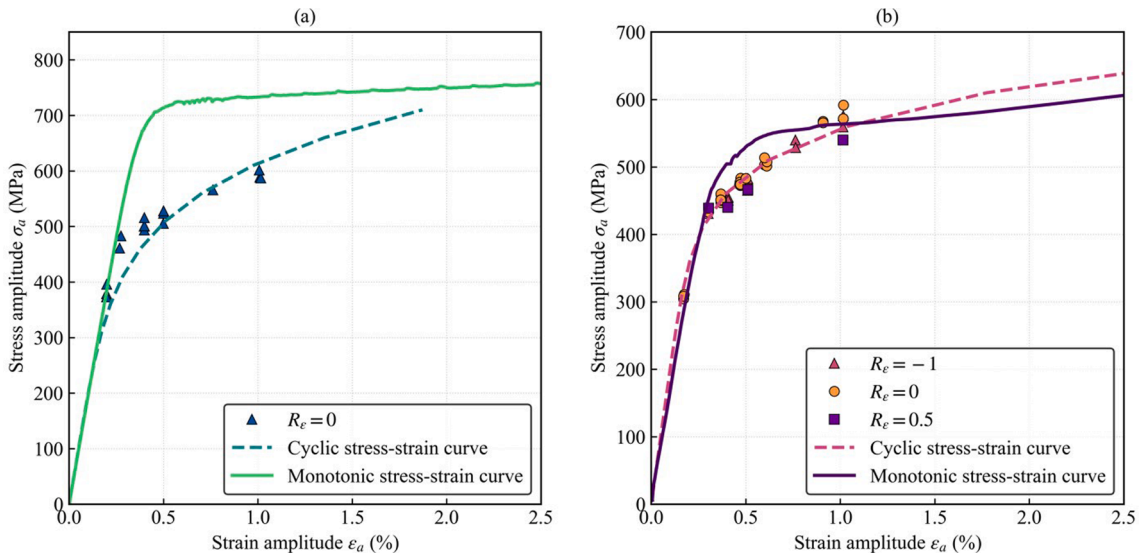


Fig. 8. Comparison between the monotonic and stabilised cyclic stress–strain responses of the WAAM-S (a) and WAAM-W (b) configurations.

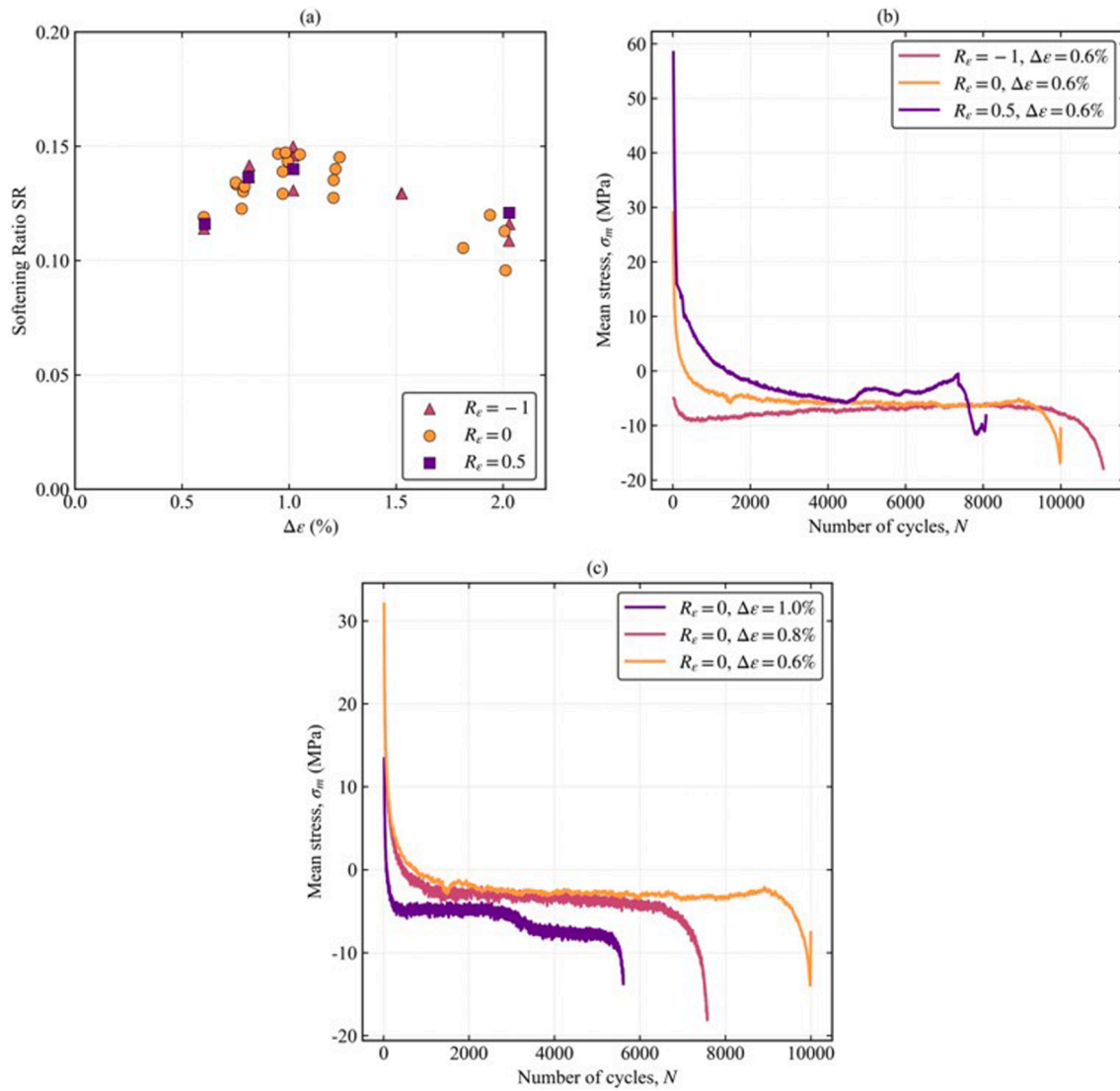


Fig. 9. Effect of strain ratio on the fatigue response of the WAAM-W configuration: (a) cyclic softening ratio versus total strain range; (b) mean stress evolution at $\Delta\epsilon = 0.6\%$ for different strain ratios; and (c) mean stress evolution at $R_\epsilon = 0$ for different strain ranges.

elastic-dominated regime, the imposed tensile mean stress relaxes at markedly different rates depending on the strain ratio, as shown in Fig. 9b. Relaxation is most pronounced for $R_\epsilon = -1$ and slowest for $R_\epsilon = 0.5$. Consequently, higher mean strain ratios lead to a greater persistence of tensile mean stress during fatigue loading, which contributes directly to the observed reduction in fatigue life. The strain dependence of this process is illustrated in Fig. 9c, where the mean stress evolution at $R_\epsilon = 0$ is shown for strain ranges of 0.6%, 0.8%, and 1.0%. Mean stress relaxation becomes increasingly pronounced with increasing strain amplitude, indicating that higher strain amplitudes promote more effective stress redistribution. This behaviour shows that mean stress relaxation is governed by the degree of cyclic plasticity developed during loading: when plastic deformation is sufficient, tensile mean stresses relax rapidly, and their influence on fatigue life is reduced; when plastic deformation is limited, mean stresses persist and exert a stronger detrimental effect. Taken together, these results show that the influence of mean strain ratio on fatigue life is controlled not by changes in the intrinsic cyclic constitutive response, but by the extent to which mean stress can relax during cyclic loading.

3.5. Microstructural observation

Fig. 10 shows that the WAAM-S and WAAM-W configurations exhibit distinct bainitic morphologies, which can be linked to their different deposition strategies. WAAM-S was produced as a multi-bead solid geometry with a higher heat input of 0.92 kJ/mm and an interpass temperature of 150 °C, whereas WAAM-W was deposited as a single-bead wall using a lower heat input of 0.51 kJ/mm and an interpass temperature of 250 °C. In WAAM-S, the larger deposited volume and overlap between adjacent beads introduced repeated local reheating and spatial variations in cooling rate, resulting in a more heterogeneous microstructure. Accordingly, WAAM-S exhibits a predominantly lath bainitic ferrite microstructure with relatively coarse packeted regions (Fig. 10a). Within this matrix, darker constituents are observed at lath intersections and packet boundaries and are interpreted as carbon-enriched regions consistent with martensite–austenite (M–A) constituents. Such microstructural heterogeneity may reduce microstructural stability under cyclic loading, consistent with the more pronounced cyclic softening observed for WAAM-S. In contrast, WAAM-W exhibits a more homogeneous bainitic microstructure, with bainitic ferrite regions that are locally more equiaxed and less pronounced packet variability, as shown in Fig. 10b.

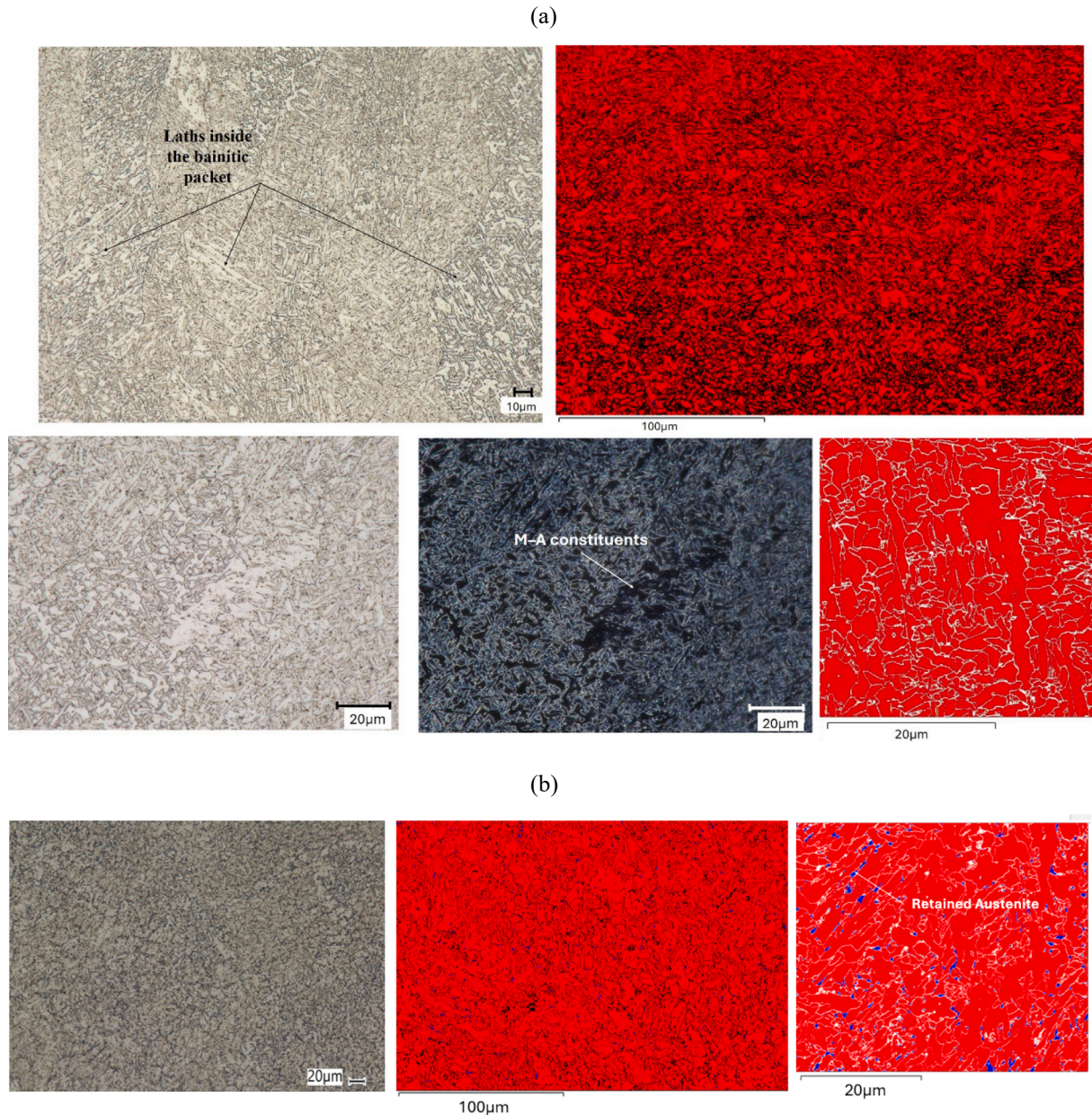


Fig. 10. (a) Optical micrographs of the WAAM-S configuration showing bainitic packets and lath structures with martensite–austenite (M–A) constituents. (b) Optical micrograph of the WAAM-W configuration and corresponding EBSD phase map indicating the presence of retained austenite.

EBSD analysis revealed the presence of retained austenite (RA) in the WAAM-W configuration, while no measurable RA was identified in WAAM-S. This observation was quantitatively confirmed by X-ray diffraction (XRD), which measured an initial retained austenite fraction of approximately 7 wt% in WAAM-W. Following cyclic loading, a systematic reduction in RA content was observed, decreasing to approximately 3.6 wt% after testing at a strain range of $\Delta\epsilon = 0.8\%$ and further to 1.4 wt% at $\Delta\epsilon = 2.0\%$. This progressive reduction provides direct evidence of strain-induced transformation of retained austenite into martensite during cyclic deformation. The extent of transformation increases with strain amplitude, indicating that the transformation is mechanically driven and coupled to cyclic plasticity. This behaviour establishes a clear quantitative link between phase evolution and the imposed cyclic strain conditions.

The measured phase transformation behaviour provides a direct explanation for the differences in cyclic softening between the two configurations. In WAAM-W, the transformation of retained austenite

during cyclic loading contributes to transformation-induced plasticity (TRIP), which enhances strain hardening and delays the rearrangement of dislocation structures [35]. As a result, the reduction in stress amplitude with cycling is moderated, leading to the lower softening ratios observed experimentally ($\sim 13\%$ at $\Delta\epsilon = 0.8\%$). In contrast, the absence of retained austenite in WAAM-S eliminates this transformation-assisted mechanism, resulting in a more rapid evolution of dislocation substructure and higher cyclic softening ($\sim 19\%$). This demonstrates that the presence and evolution of retained austenite are directly coupled to cyclic plasticity and cannot be treated as a separate microstructural observation.

The influence of retained austenite extends beyond cyclic stress response and directly affects fatigue performance. The TRIP-assisted deformation mechanism in WAAM-W promotes more distributed plasticity and reduces early strain localisation, thereby delaying fatigue crack initiation and contributing to the higher fatigue ductility coefficient and longer fatigue lives observed in this configuration. The

quantitative correlation between decreasing retained austenite fraction, increasing strain amplitude, and cyclic softening behaviour provides a consistent and experimentally supported framework linking phase transformation to fatigue performance.

The combined EBSD and XRD analyses demonstrate that the interaction between microstructural heterogeneity, crystallographic orientation distribution, and strain-induced phase transformation governs cyclic deformation and fatigue performance in WAAM-fabricated high-strength steel. In particular, the presence and evolution of retained austenite provide a measurable and mechanistically supported link between microstructure and cyclic softening, thereby bridging the gap between mechanical response and microstructural characterisation.

The EBSD inverse pole figure (IPF) maps shown in Fig. 11a further highlight the morphological differences between the two configurations. The WAAM-S material exhibits locally aligned and elongated bainitic features forming banded regions, whereas the WAAM-W configuration displays a more homogeneous orientation distribution with less pronounced alignment of bainitic laths. To further quantify the average grain size, EBSD analysis based on the equivalent circle diameter (ECD) was performed. As shown in Fig. 11b, WAAM-S exhibits a smaller mean equivalent circle diameter ($1.37 \pm 1.17 \mu\text{m}$) compared with WAAM-W

($1.81 \pm 1.80 \mu\text{m}$), indicating a finer crystallographic substructure in WAAM-S.

The EBSD-derived misorientation and orientation maps in Fig. 11c provide a direct link between the measured cyclic deformation behaviour and the underlying microstructure. The WAAM-S configuration exhibited a slightly higher average KAM value of approximately 0.42° , compared with 0.34° for WAAM-W, suggesting a modestly higher level of dislocation density. Although the difference in the average KAM values is relatively small, the spatial distribution of the misorientation provides a clearer distinction between the two configurations. In the WAAM-S configuration, the presence of banded regions with elevated misorientation and clustered WBV intensity indicates the formation of geometrically necessary dislocation (GND) networks concentrated along preferential paths. These regions act as localised deformation zones under cyclic loading, facilitating early strain localisation and accelerating cyclic softening. This behaviour is consistent with the experimentally observed higher plastic strain ranges and softening ratios for WAAM-S. In contrast, the WAAM-W configuration exhibits a more spatially dispersed distribution of misorientation, indicating that GND accumulation is more uniformly distributed. This promotes a more stable cyclic response by delaying the formation of dominant localisation

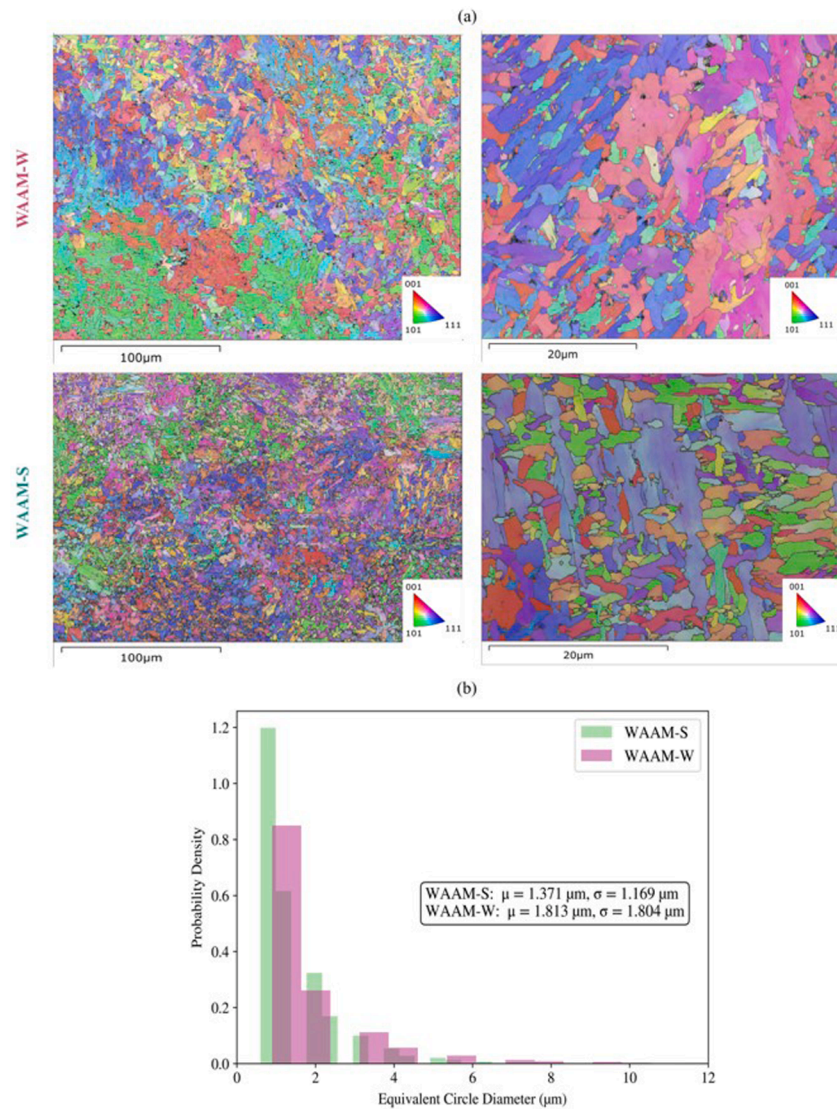


Fig. 11. (a) EBSD inverse pole figure (IPF) orientation maps of the WAAM-W and WAAM-S configurations. (b) Grain size distribution expressed as equivalent circle diameter (ECD) for both configurations. (c) Weighted Burgers Vector (WBV) maps in sample coordinates showing the spatial distribution of geometrically necessary dislocations and Kernel Average Misorientation (KAM) maps illustrating local misorientation.

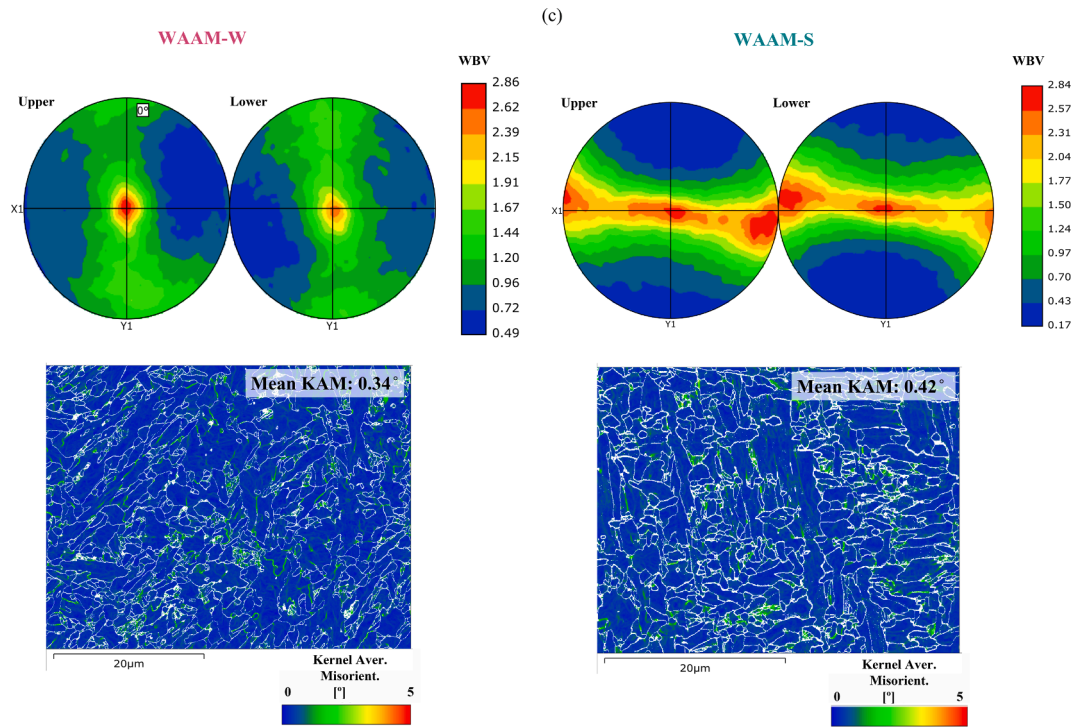


Fig. 11. (continued).

bands and enabling more homogeneous cyclic plasticity. As a result, WAAM-W maintains lower cyclic softening and improved fatigue resistance. These observations confirm that the spatial distribution of misorientation and texture plays a critical role in governing cyclic deformation mechanisms in WAAM steels. The observed differences in cyclic softening behaviour between the WAAM-S and WAAM-W configurations, discussed in section 3.2, can be directly interpreted in terms of their underlying microstructural and crystallographic characteristics. The WAAM-S configuration, which exhibits higher softening ratios and larger stabilised plastic strain ranges, is characterised by a heterogeneous bainitic microstructure with locally aligned and banded regions, as revealed by EBSD IPF and WBV maps. These features promote non-uniform strain partitioning and the development of local strain incompatibilities during cyclic loading, which accelerate dislocation rearrangement and reduce the internal resistance to plastic deformation. Consequently, cyclic softening proceeds more rapidly in WAAM-S despite its higher initial strength. In contrast, the WAAM-W configuration shows a more homogeneous orientation distribution and a less pronounced alignment of bainitic laths. This results in a more uniform accommodation of plastic strain at the microstructural level, reducing local strain concentrations and delaying microstructural degradation. This difference is directly reflected in the lower cyclic softening ratios and smaller plastic strain accumulation observed experimentally. These results demonstrate that cyclic plasticity in WAAM-fabricated steel is governed not only by strength level but also by the spatial distribution of crystallographic orientations and the degree of microstructural heterogeneity.

3.6. Fractography

Fractographic analysis of representative specimens from both configurations, shown in Fig. 12, confirms that fatigue failure was governed by surface crack initiation, subsequent stable crack propagation, and final ductile overload. Most fatigue crack nucleation sites were located at or near the specimen surface along the free edges of the gauge section, consistent with the larger inelastic deformation experienced in this region and the development of surface intrusion–extrusion features during

cyclic loading. Multiple crack initiation sites were frequently observed, indicating distributed cyclic strain localisation rather than failure controlled by a single dominant defect. The fatigue crack propagation region was characterised by directional growth features and a curved crack front advancing progressively across the specimen cross-section, followed by a transition to a rougher final fracture zone associated with rapid failure of the remaining ligament.

For the WAAM-W configuration, differences in crack development were observed at a total strain range of $\Delta\epsilon = 0.6\%$, where specimens tested at $R_e = -1$ exhibited a more extended fatigue crack propagation region and a relatively smoother crack growth surface, reflecting more gradual crack advance over a larger number of cycles. In contrast, specimens tested at $R_e = 0.5$ showed a shorter propagation region and a more abrupt transition to final fracture. At higher strain levels, specifically at $\Delta\epsilon = 0.75\%$ and $R_e = 0$, the fracture surfaces displayed a reduced propagation region and a more pronounced final fracture zone.

For the WAAM-S configuration, specimens tested at $\Delta\epsilon = 0.8\%$ in both horizontal and vertical orientations exhibited a similar fracture sequence, with crack initiation occurring predominantly at the specimen surface and propagation characterised by a curved crack front advancing across the cross-section. Compared with WAAM-W configuration at comparable strain levels, WAAM-S specimens generally exhibited a relatively shorter fatigue crack propagation region and a more dominant final fracture zone. At higher strain levels ($\Delta\epsilon = 1\text{--}2\%$), the final fracture region became more pronounced, indicating accelerated crack propagation with increasing cyclic plastic deformation. Higher-magnification observations of the final fracture regions revealed a dense distribution of dimples and voids in both configurations, indicating ductile fracture through micro-void nucleation, growth, and coalescence.

4. Fatigue Life Prediction Models

Fatigue-life relationships describe the link between fatigue life, typically represented as the number of reversals to failure $2N_f$, and a fatigue damage parameter. These relationships are widely used in engineering design to account for fatigue damage in structural components

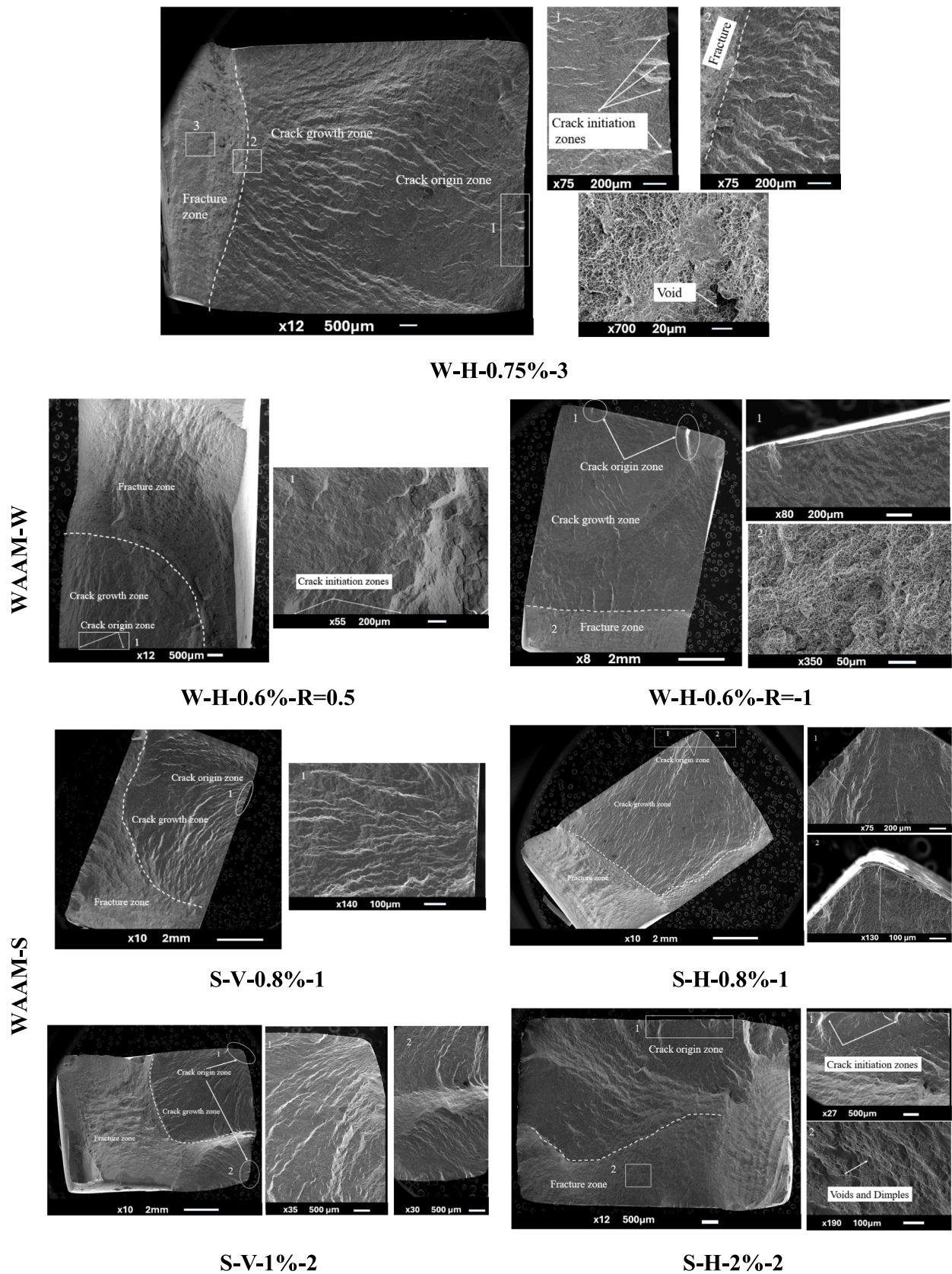


Fig. 12. Fractographic features of representative WAAM-W and WAAM-S specimens after low-cycle fatigue testing at different strain ranges and strain ratios.

and materials. In the present study, three models are considered to describe the fatigue behaviour of WAAM-W configuration: the Basquin-Coffin-Manson relationships, the Smith-Watson-Topper (SWT) model, and the Total Strain Energy Density (TSED) model. The Basquin-Coffin-Manson and TSED approaches are applied to both WAAM-W and WAAM-S, whereas the SWT model is evaluated only for the WAAM-W configuration, since fully reversed strain-controlled fatigue tests ($R_e = -1$) were performed only for the WAAM-W specimens.

4.1. Strain-life behaviour

The low-cycle fatigue behaviour of WAAM-W and WAAM-S was evaluated using the total strain-life approach, in which the elastic and plastic strain components are described by the Basquin and Coffin-Manson relationships, respectively, as given by Eq. (4). The fitted parameters are summarised in Table 7, and the corresponding strain-life curves are shown in Fig. 8.

$$\frac{\Delta \epsilon_t}{2} = \frac{\Delta \epsilon_e}{2} + \frac{\Delta \epsilon_p}{2} = \frac{\sigma'_f}{E} (2N_f)^b + \epsilon'_f (2N_f)^c \quad (4)$$

A clear divergence in fatigue response is observed between the WAAM-W and WAAM-S configurations, as shown in Fig. 13. This difference is governed by the contrast between fatigue ductility and fatigue strength. The WAAM-W configuration exhibits a substantially higher fatigue ductility coefficient ($\epsilon'_f = 0.650$) than the WAAM-S configuration ($\epsilon'_f = 0.138$), indicating a much greater capacity for cyclic plastic strain accommodation prior to failure. The WAAM-S configuration exhibits a significantly higher fatigue strength coefficient ($\sigma'_f = 2584.4$ MPa) than WAAM-W ($\sigma'_f = 1081.2$ MPa), consistent with its higher initial strength level and lower ductility observed in the monotonic tensile response. The differences in strain-life behaviour between the two configurations can be directly linked to the previously discussed microstructural characteristics. The improved fatigue ductility of WAAM-W is associated with its more homogeneous bainitic structure and distributed deformation mechanisms, which allow for greater plastic strain accommodation prior to crack initiation. In contrast, the heterogeneous and banded microstructure of WAAM-S promotes local strain concentration, leading to earlier microstructural degradation and a steeper reduction in fatigue resistance with increasing number of cycles. This establishes a direct structure-property relationship between WAAM-induced microstructural features and the observed fatigue performance.

In the low-life regime, the higher ϵ'_f of WAAM-W allows larger plastic strain amplitudes to be sustained for a given fatigue life, reflecting a more deformation-tolerant response. At longer fatigue lives, however, the WAAM-S configuration exhibits higher plastic strain amplitudes than the WAAM-W for an equivalent life. This inversion reflects the stronger cyclic degradation of WAAM-S, which progressively increases the plastic strain component of the imposed total strain.

The elastic strain-life response reinforces this interpretation. Although the higher σ'_f of WAAM-S reflects its greater initial resistance to deformation, the steeper Basquin exponent ($b = -0.21$, compared with -0.086 for WAAM-W) indicates a more rapid loss of fatigue resistance with increasing life. This is further reflected in the higher transition life of the WAAM-S configuration, defined as the intersection between the elastic and plastic strain-life components, where $\Delta \epsilon_e/2 = \Delta \epsilon_p/2$. The transition life is $2N_T = 9331$ reversals for WAAM-S, compared with $2N_T = 3360$ reversals for WAAM-W, indicating that

Table 7
Summary of fatigue parameters.

Material	σ'_f	b	ϵ'_f	c	$2N_T$
WAAM-W	1081.2	-0.086	0.668	-0.684	3360
WAAM-S	2584.4	-0.21	0.138	-0.468	9331

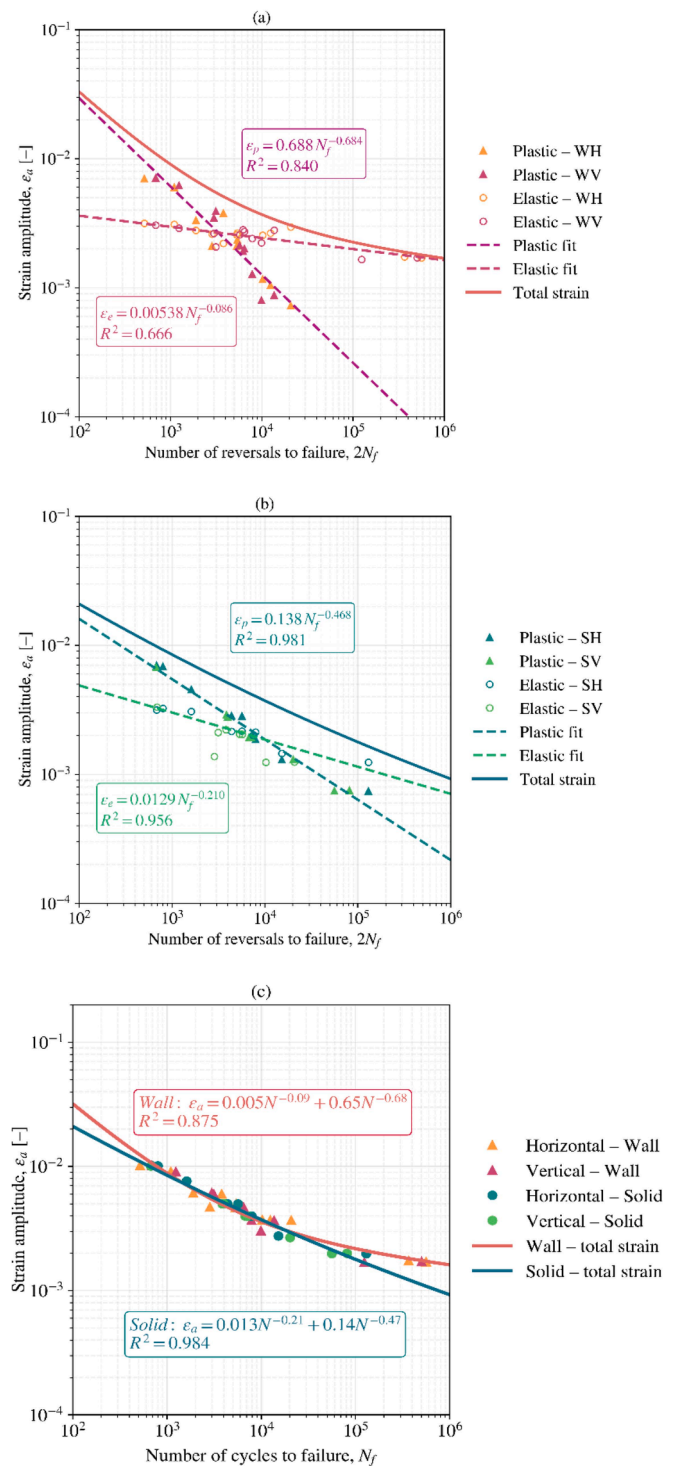


Fig. 13. Fitted elastic, plastic, and total strain-life curves for the WAAM-W and WAAM-S configurations at $R_e = 0$

elastic-plastic interaction remains important over a broader fatigue-life range in the WAAM-S configuration. These results show that the superior fatigue performance of WAAM-W arises from greater plastic strain tolerance, whereas the higher initial strength of WAAM-S does not translate into improved strain-life resistance.

Fig. 14 compares the strain-life responses of WAAM-W and WAAM-S with those of conventional S690 steel [31] and WAAM ER70S-6 and high-strength ER110S-G steels reported in the literature [24,26]. In the low-cycle regime, WAAM-W exhibits a high strain-accommodation

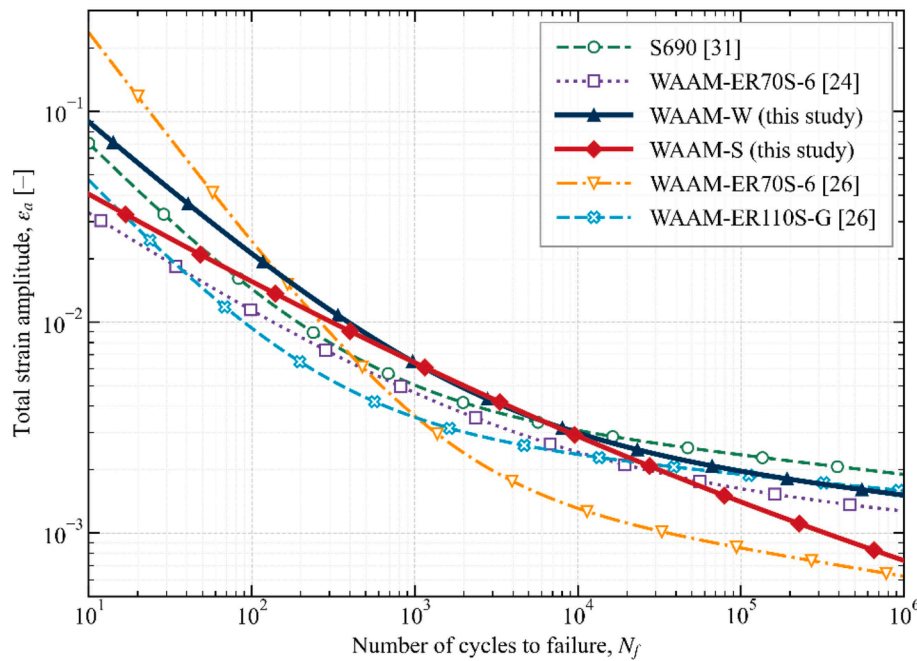


Fig. 14. Strain-life curves derived from the Coffin–Manson–Basquin relationship comparing WAAM-W and WAAM-S configurations from this study with conventional S690 steel and WAAM-ER70 steel reported in the literature.

capacity, comparable to S690 and generally above the ER70S-6 response reported in [24]. The ER70S-6 material reported in [26] reaches higher strain amplitudes at very short lives, reflecting its large fatigue-ductility contribution; however, its strain-life curve decreases more rapidly and falls below that of WAAM-W as fatigue life increases. WAAM-W therefore provides a favourable balance between plastic-strain accommodation in the short-life regime and retention of strain capacity toward longer lives. In contrast, WAAM-S exhibits the steepest strain-life response and progressively falls below the other materials with increasing fatigue life. A similarly rapid decline is observed for the high-strength ER110S-G material reported in [26], which also reaches comparatively low strain amplitudes in the longer-life regime. The comparable trends of WAAM-S and ER110S-G suggest that their greater cyclic strength is accompanied by more limited fatigue ductility, although differences in deposition strategy, thermal history, and microstructure must also be considered. Overall, the variation among the WAAM datasets demonstrates that strain-life behaviour is strongly dependent on the manufacturing route and the resulting balance between cyclic strength, microstructural stability, and plastic-strain capacity.

Furthermore, separate Basquin–Coffin–Manson parameters were fitted for the additional strain ratios in WAAM-W. The stabilised stress-strain characteristics and corresponding fatigue lives obtained at $R_\epsilon = -1$ and $R_\epsilon = 0.5$ are summarised in Table 8, while the fitted strain-life fatigue parameters are listed in Table 9. The results show a clear regime-dependent sensitivity to the mean strain ratio. As shown in Fig. 15, the strain-life curves largely overlap at higher strain amplitudes ($\frac{\Delta\epsilon}{2} \geq 0.55\%$), where the response is dominated by plastic deformation. At lower strain amplitudes, however, a systematic separation develops, with fatigue life decreasing as the mean strain ratio increases. The fitted low-cycle fatigue parameters for the different strain ratios are listed in Table 9. These results indicate that the effect of mean strain is negligible when cyclic plasticity is sufficient to redistribute the imposed stress state, yet becomes significant when deformation is more strongly constrained by elastic response.

Table 8

Stabilised stress-strain characteristics and fatigue lives of the WAAM-W specimens tested at different mean strain ratios.

Specimen	$\Delta\sigma$ [MPa]	σ_m [MPa]	$\Delta\epsilon$ [%]	$\Delta\epsilon_p$ [%]	$\Delta\epsilon_e$ [%]	N_f [Cycles]
W-H-2%-R = -1	1151.6	-11.5	2.03	1.47	0.56	720
W-H-2%-R = 0.5	1228.4	-8.4	2.03	1.32	0.71	1059
W-H-1.5%-R = -1	1057.3	-10.4	1.53	0.98	0.55	1605
W-H-1%-R = -1	965.6	-9.2	1.02	0.52	0.50	4352
W-H-1%-R = 0.5	1084.7	-7.6	1.02	0.51	0.51	4083
W-H-0.8%-R = -1	901.5	-6.3	0.81	0.34	0.47	7465
W-H-0.8%-R = 0.5	1019.4	-3.5	0.81	0.32	0.49	5825
W-H-0.6%-R = 0.5	993.3	-12.3	0.61	0.21	0.40	8069
W-V-2%-R = -1	1118.5	-11.8	2.03	1.42	0.61	718
W-V-1.5%-R = -1	1079.7	-10.3	1.53	0.99	0.54	1280
W-V-1%-R = -1	930.9	-7.6	1.02	0.50	0.52	3579
W-V-0.8%-R = -1	910.1	-6.8	0.81	0.33	0.48	7134
W-V-0.6%-R = -1	861.8	-11.9	0.60	0.13	0.47	11,344

Table 9

Strain-life fatigue parameters for WAAM-W configuration obtained at different mean strain ratios.

WAAM-W	σ'_f	b	ϵ'_f	c	$2N_T$
$R_\epsilon = -1$	985.3	0.080	0.998	-0.735	3340
$R_\epsilon = 0$	1081.2	-0.086	0.668	-0.684	3360
$R_\epsilon = 0.5$	2369.6	-0.266	3.03	-0.872	3161

4.2. Smith–Watson–Topper (SWT) model

The Smith–Watson–Topper (SWT) model introduces a strain-energy-based fatigue damage parameter to account for the influence of mean stress on fatigue life. The SWT parameter combines the maximum stress and strain amplitude within a cycle and is expressed as:

$$\sigma_{max}\epsilon_a = \frac{(\sigma'_f)^2}{E}(2N_f)^{2b} + \sigma'_f\epsilon'_f(2N_f)^{b+c} \quad (5)$$

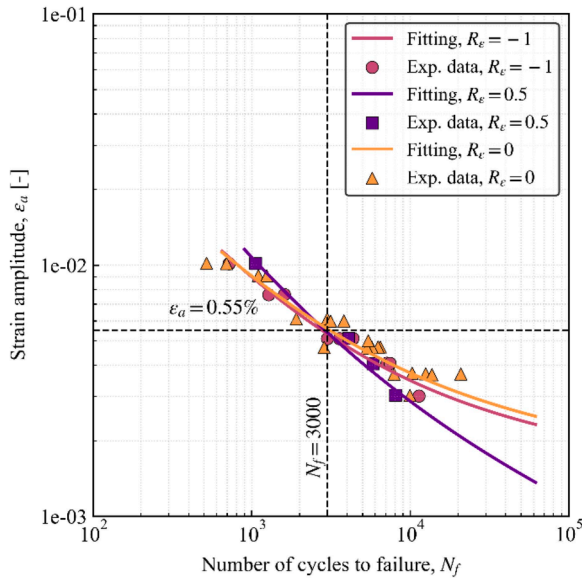


Fig. 15. Strain-life curves for $R_e = -1, 0$, and 0.5 .

where σ_{max} is the maximum stress and ε_a is the strain amplitude. The material parameters $\sigma'_f, \varepsilon'_f, b$ and c were obtained from the Basquin and Coffin-Manson strain-life relationship derived from the fully reversed ($R = -1$) strain-controlled fatigue tests. The SWT relation was subsequently used to predict the fatigue life of the WAAM-W specimens under different mean strain ratios.

Fig. 16 compares the predicted fatigue lives with the corresponding experimental results. The data points are generally close to the unity line, indicating satisfactory agreement between the SWT predictions and the experimental fatigue lives. The scatter relative to the unity line was quantified using the standard deviation of $\log_{10}(N_{pred}/N_{exp})$, yielding $S = 0.146$. This corresponds to a $\pm 2S$ scatter factor of approximately $10^{0.25} = 1.96$. The mean logarithmic error is -0.0215 , corresponding to an average prediction ratio of $N_{pred}/N_{exp} \approx 0.95$, which indicates a slight conservative bias. The predictions show no clear systematic dependence on the investigated strain ratios ($R_e = -1, 0$ and 0.5), consistent with the relatively small measured mean stresses after cyclic relaxation. Slightly

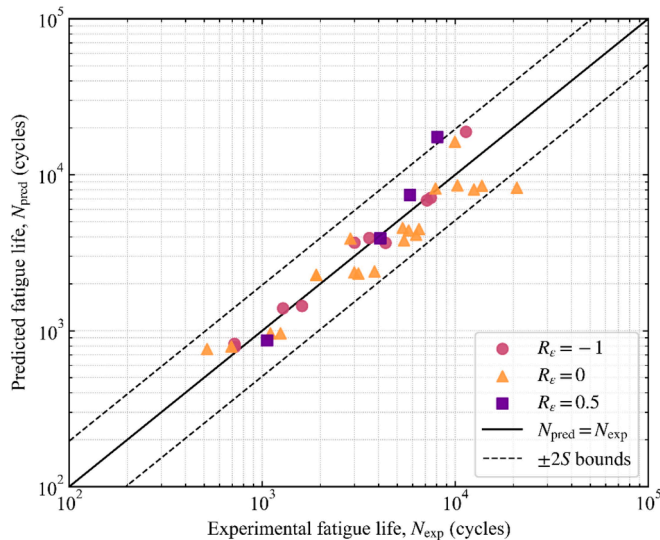


Fig. 16. Comparison between predicted fatigue life and experimentally measured fatigue life for WAAM-W configuration using the Smith-Watson-Topper (SWT) model.

larger deviations are observed in the higher-life regime (10^4 - 10^5 cycles), where some predictions diverge from the unity line.

4.3. Total Strain Energy Density (TSED) model

The strain energy density approach was considered as an alternative approach to predict the fatigue life. The fatigue damage is assumed to be governed by the energy dissipated during cyclic deformation. The plastic strain energy density per cycle, ΔW_p , corresponds to the area enclosed by the stabilized stress-strain hysteresis loop and represents the plastic work dissipated during cyclic loading. The relationship between the plastic strain energy density and the number of reversals to failure can be described by Eq. (6):

$$\Delta W_p = \kappa_p (N_f)^{\alpha_p} \quad (6)$$

where κ_p and α_p are material parameters determined from regression of the experimental data. Although the plastic strain energy density provides a suitable description of fatigue damage in the low-cycle fatigue regime, its applicability becomes limited at longer fatigue lives, where the cyclic response is increasingly governed by elastic deformation. To account for this contribution and the effect of tensile mean stress, the total strain energy density per cycle was therefore considered as the governing damage parameter. The total strain energy density is defined by Eq. (7) as the sum of the plastic strain energy density and the positive elastic strain energy density associated with the tensile portion of the loading cycle:

$$\Delta W_t = \Delta W_p + \Delta W_e \quad (7)$$

where the positive elastic strain energy density ΔW_e was calculated from the maximum tensile stress of the stabilized cycle.

Fig. 17 illustrates the evolution of the plastic, elastic, and total strain energy densities as a function of the number of cycles to failure over the investigated strain range of 2%–0.4%. For both configurations, the plastic strain energy density exhibits a steeper dependence on fatigue life than the elastic component, indicating that cyclic plastic deformation dominates the energy dissipation in the low-cycle fatigue regime. For the WAAM-S configuration, the slopes of the plastic and elastic components are approximately -0.58 and -0.31 , respectively, resulting in the total strain energy density closely following the plastic strain energy trend over most of the investigated fatigue lives, see Fig. 17a. In contrast, a stronger separation between the plastic and total strain energy curves is observed for the WAAM-W configuration, as shown in Fig. 17b, indicating a larger relative contribution of the elastic strain energy component at longer fatigue lives. Furthermore, Fig. 17c shows that the WAAM-W configuration exhibits lower total strain energy density values for a given fatigue life compared to the WAAM-S configuration, indicating that less cyclic energy is dissipated per cycle in WAAM-W.

The influence of the strain ratio on the fatigue behavior was further evaluated using the total strain energy density approach. Fig. 18 shows the relationship between the total strain energy density and the number of cycles to failure for the investigated strain ratios $R = -1$, $R = 0$, and $R = 0.5$. In all cases, the total strain energy density decreases with increasing fatigue life following a power-law relationship. A clear influence of the strain ratio is observed, where higher mean stress conditions (e.g., $R = 0.5$) result in larger total strain energy density values for a given fatigue life compared to fully reversed loading ($R = -1$). The fitted curves show good agreement with the experimental data, as reflected by the high correlation coefficients. These results indicate that the strain ratio significantly affects the cyclic energy dissipation and, therefore, the fatigue response of the material.

The constants associated with each energy component are summarized in Fig. 18, together with the corresponding correlation coefficients for both configurations and strain ratios.

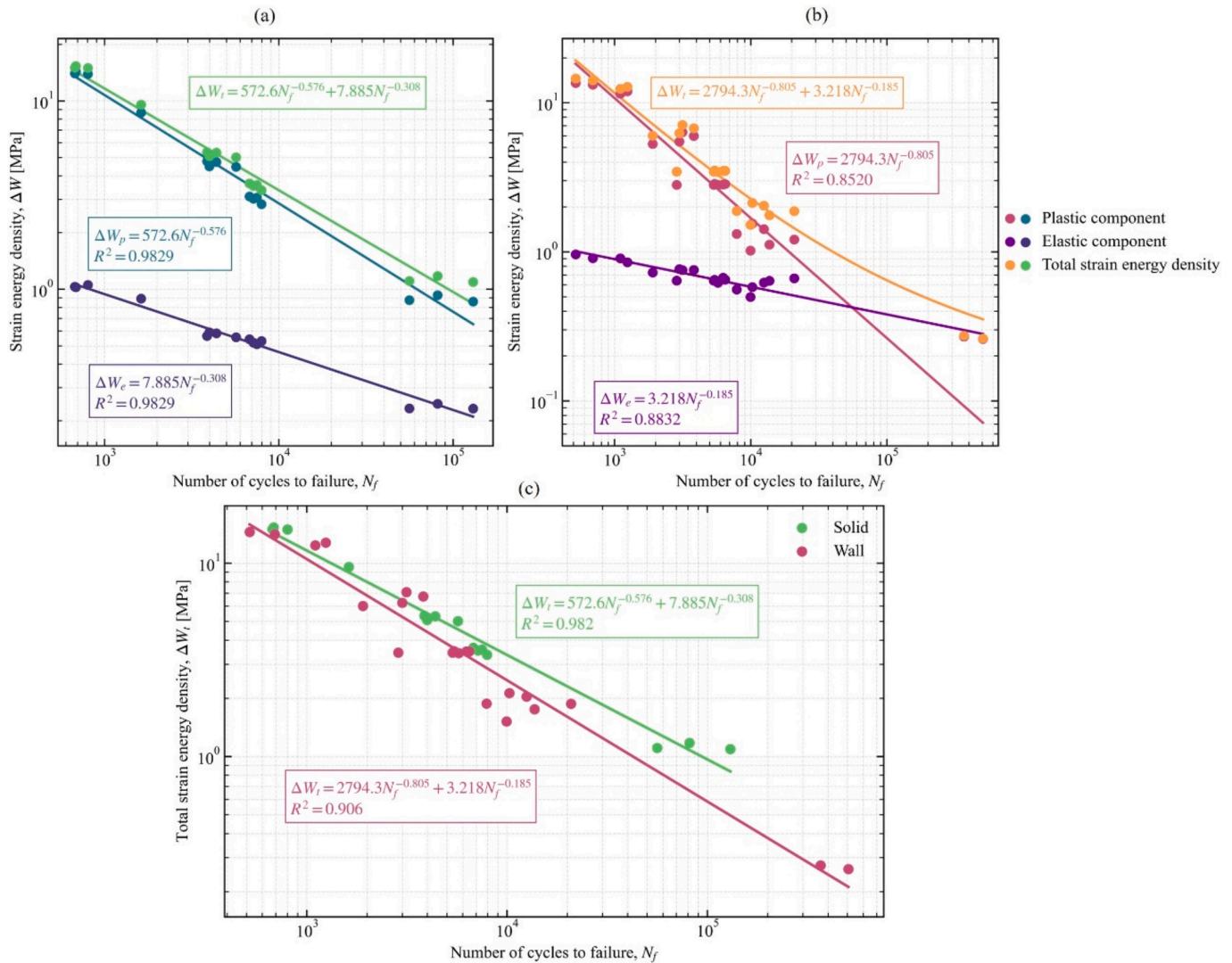


Fig. 17. Plastic, elastic, and total strain energy density as a function of the number of cycles to failure for the (a) WAAM-S and (b) WAAM-W, (c) together with the comparison of the total strain energy density for both configurations.

5. Conclusions

The low-cycle fatigue behaviour of WAAM-fabricated AM70 high-strength steel is investigated experimentally and analytically for thin-walled and solid deposits produced under different WAAM thermal histories. Strain-controlled fatigue testing, cyclic stress-strain analysis, microstructural characterisation, fractography and fatigue-life modelling lead to the following conclusions:

1. WAAM-S exhibited higher monotonic and stabilised cyclic strength than WAAM-W, but lower cyclic stability. The 0.2% offset cyclic yield stress decreased by approximately 35% for WAAM-S, compared with about 17% for WAAM-W, demonstrating a clear trade-off between cyclic strength and resistance to cyclic softening.
2. The stabilised hysteresis loops and cyclic stress-strain curves show limited orientation dependence, indicating negligible cyclic anisotropy and a relatively uniform cyclic response in both WAAM-W and WAAM-S.
3. For the investigated WAAM-W specimens, mean strain effects are most pronounced in the lower strain-amplitude regime, where limited cyclic plasticity leads to persistent tensile mean stress and reduced fatigue life under positive mean strain. At higher strain amplitudes, more extensive cyclic plasticity promotes rapid mean stress relaxation, causing the strain-life responses for different strain ratios to converge. Since the variable mean-strain tests were based on one specimen per condition and were not conducted for WAAM-S, these observations are indicative trends specific to the investigated thin-walled configuration.
4. The different WAAM thermal histories result in distinct microstructural conditions. WAAM-S exhibits a coarser and more heterogeneous bainitic microstructure with locally concentrated misorientation, which is consistent with stronger cyclic softening and increased cyclic plastic strain accumulation. In contrast, WAAM-W shows a more homogeneous bainitic microstructure and retained austenite, supporting more distributed cyclic deformation.
5. The retained austenite content in WAAM-W decreased progressively with increasing applied strain range, providing evidence of strain-induced transformation during fatigue. This transformation-assisted mechanism contributes to strain hardening and distributed plasticity, thereby moderating cyclic softening and improving cyclic stability in the thin-walled configuration.
6. The strain-life response confirms a clear configuration dependence. WAAM-W exhibits higher fatigue resistance in the low-cycle fatigue regime, whereas WAAM-S shows a steeper strain-life slope and becomes less favourable at longer fatigue lives.

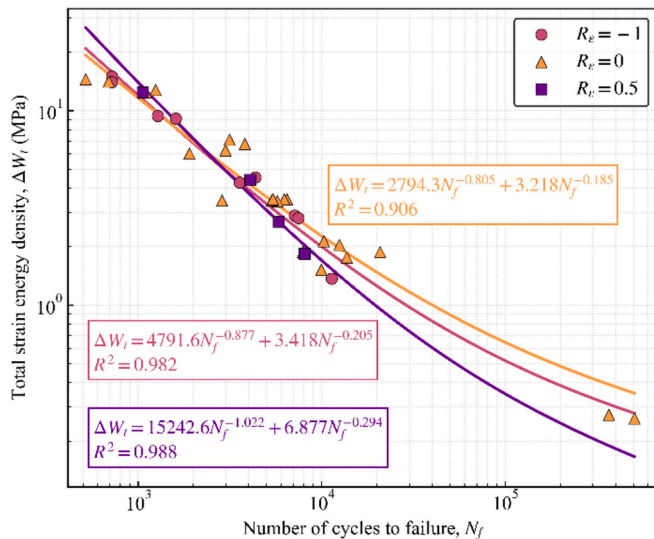


Fig. 18. Total strain energy density ΔW_f as a function of the number of cycles to failure N_f for strain ratios $R = -1$, $R = 0$, and $R = 0.5$. Symbols represent experimental data, and solid lines represent the fitted relations obtained from the plastic and elastic strain energy components.

7. For WAAM-W, the SWT formulation provided a unified fatigue-life correlation across the investigated strain ratios by accounting for the effect of tensile mean stress. The energy-based analysis related the fatigue response of both configurations to the elastic and plastic strain-energy contributions, with the larger plastic-energy dissipation of WAAM-S being consistent with its stronger cyclic softening

The results quantify the fatigue behaviour of WAAM-fabricated AM70 steel, which is governed by the interaction between cyclic softening, mean-stress relaxation, plastic-strain capacity, and microstructural stability. For structural fatigue prediction, the deposit configuration and its associated thermal history should be considered explicitly, rather than treating WAAM high-strength steel as a single homogeneous material.

CRediT authorship contribution statement

Hagar El Bamby: Writing – review & editing, Writing – original draft, Visualization, Validation, Methodology, Investigation, Formal analysis. **Vera Popovich:** Validation, Investigation, Conceptualization. **Ramesh Babu:** Writing – review & editing, Validation. **Milan Veljkovic:** Writing – review & editing, Supervision. **Trayana Tankova:** Writing – review & editing, Validation, Methodology, Funding acquisition, Formal analysis, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Declaration of AI-assisted technologies in the writing process

The authors used generative artificial intelligence (AI) tools to assist with language editing during the preparation of this manuscript and in the generation of Figs. 1 and 2. All scientific analyses, interpretations, and conclusions were performed by the authors, who take full responsibility for the content of the manuscript.

Data availability

Data will be made available on request.

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