SOME ASPECTS OF THE ENVIRONMENTAL EFFECT ON THE FATIGUE MECHANISM OF A HIGH STRENGTH ALUMINIUM ALLOY

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SUMMARY
Theoretical studies on crack growth usually consider the behaviour of the material (plasticity) and the state of stress only. It is the aim of this paper to show that the environment essentially change the mechanism of crack growth.
Crack propagation tests were conducted using Aluminium 7075-T6 sheet to study the relationship between the crack propagation rate and the stress intensity range in different environments. The crack propagation results are complemented by fractographic analysis.
The author strongly supports a model based on ideas as postulated by TOMKINS, LAIRD and KELLY. The significance for crack growth under corrosion fatigue conditions is discussed.
1. INTRODUCTION

The most important parameters affecting fatigue crack growth are:
- material properties
- state of cyclic stress at the crack tip
- environment at the crack tip

Theoretical studies on crack growth usually consider only the behaviour of the material (plasticity) and the state of stress. With respect to the environment it is generally recognised that an agressive environment will accelerate crack growth. However, the environment may be doing more: it can essentially change the mechanism of crack growth. This is the main theme of this paper. It will be supported by new fractographic observations on crack growth in 7075 - T6 aluminium alloy sheet material.

The paper consists of three parts:
1. Definitions of the transition from tensile mode to shear mode and the transition of the state of stress from plane strain to plane stress. In the literature both transitions are usually supposed to be identical. It will be shown that this is incorrect, and that it can only be understood by introducing environmental aspects into the fatigue crack mechanism.
2. Relevant empirical evidence will be summarized. This will include results from recent tests on the effects of environment, applied potential and temperature.
3. Analysis of the crack growth mechanism. Aspects of electrochemistry in relation to decohesion will be introduced. A model will be presented, starting from ideas postulated by TOMKINS [1], LAIRD [2] and KELLY et al [3], but including electrochemical features.
2. THE TRANSITION FROM TENSILE MODE TO SHEAR MODE

The fracture surfaces of centre-cracked sheet specimens subjected to cyclic loading in air show characteristic features (Figure 1). Initially crack growth occurs macroscopically in the so-called tensile mode on a plane perpendicular to the loading direction. As the crack length increases or at higher ΔK values, cracking occurs in the shear mode on a plane inclined at 45 degrees to the loading direction. Transition from the tensile mode to the shear mode occurs gradually, starting with shear lips at the two surfaces (Figure 2).

Many investigators suggest that during the tensile mode of cracking the state of stress is triaxial (plane strain), and during the shear mode of cracking the state of stress is biaxial (plane stress). Tests with specimens from different thicknesses seem to confirm this statement (Figure 3). The crack length at which the transition is completed is strongly dependent on the sheet thickness. For greater sheet thicknesses but with the same applied load, the transition occurs at higher ΔK-values. The effect of the stress, $\sigma_z$, in the thickness direction is clear.

The above evidence suggests that a state of plane strain will produce a tensile mode crack, and a state of plane stress, a shear mode crack. However, this evidence was obtained in tests carried out in normal air. It will be shown later that a more aggressive environment (salt water) can induce a tensile mode crack at ΔK-values producing shear mode cracks in normal air. Figure 4 shows the fracture surfaces of three specimens fatigued in vacuum, normal air and salt water. It is clearly shown that the tensile mode area is greater the more aggressive the environment. These results have led the author to the conclusion that a physical correlation between the state of stress and the mode of cracking does not exist. Apparently, the environment has a significant effect on the mode of cracking.
3. SUMMARY OF EMPIRICAL EVIDENCE

In the present investigation fatigue crack propagation tests in different environments have been carried out on centre cracked sheet specimens. All experiments were made in tension-tension loading in an Amsler electro-hydraulic system. A survey of the tests is given in Table 1. Figure 5 clearly shows the environmental effects on the fatigue life. The difference in crack propagation speed is larger at low ΔK-values (Figure 6). The reasons for this behaviour may be:

(i) low crack propagation speed, which implies that more time is available for electrochemical reactions

(ii) state of stress near the crack tip (plane strain)

Successively a number of experiments from Table 1 will be discussed shortly in this paper.

Inert environments:
Tests were carried out in high vacuum, dry nitrogen, dry oxygen and a low vacuum. In these environments the fatigue life is larger than in air. It is striking to note that the fatigue life in dry oxygen does not differ much from that in high vacuum. Water vapour must be the cause of the large differences between the fatigue life in normal air and in inert environments. The macroscopic fracture surfaces in inert environments appear to be much rougher than those in humid environments (Figures 7 and 8). It may even be wondered if the fracture started in the tensile mode at all. It might be better to speak about multiple shear. Figures 9 and 10 show this behaviour more clearly. These photographs show cross-sections at low ΔK-values from fracture surfaces in vacuum and in air respectively. It appears that crack extension in an inert environment is attended with more plastic deformation than in aggressive environments.

Another observation of the macroscopic fracture surface is associated with the transition from the tensile mode to the shear mode. In air the transition occurs at a later stage (higher ΔK-values) than in inert environments (Figure 3). In the past the transition was frequently related to ductility alone. It now appears that the environment is also highly
significant for this phenomenon.
In the literature a total absence of striations on fracture surfaces in vacuum is generally reported (MEYN [4], PELLOUX [5], BROEK [6]). In the present investigation however striations were found on the 7075-T6 fracture surfaces in vacuum (Figure 11 and 12). These striations are less regular than in air (Figure 13), which again indicated more intensive plastic deformation without cleavage. Striations in vacuum tests on Al-alloys were also reported by WANNILL [7].

Salt water:
A 3.5 pct. aqueous NaCl environment lowers considerably the life of a fatigue specimen. Strong pitting occurs of the highly stressed material at the crack tip.
The temperature has an important influence on electro-chemical reactions. Lowering the temperature implies a retardation of the corrosion reactions i.e. a lower crack propagation rate (Figure 14).
It is interesting to note that the tensile mode area is larger at higher temperature, which confirms the previous statement: the environment has a significant effect on the mode of cracking.

Buffer-inhibitor solution / Applied potentials:
An effective method to study environmental effects is by means of applied potentials. The environment used was aerated 0.05 molal potassium hydrogen phthalate and 0.2 molal potassium sulphate in distilled water; pH was 4.0. This buffer-inhibitor solution was chosen because of its non aggressive character and its good conductivity. The fatigue crack growth rates in this solution reduces from that observed in distilled water to that observed in laboratory air.
Inhibition by ion adsorption and buffering from the electrolyte at the crack tip (constant pH) might be the reason.
Some data obtained by polarization studies shown in Figure 15 clearly indicate that at growing anodic potentials the fatigue life decreases. A similar behaviour was found for potentials below 0.3 Volts (S.H.E.). An
explanation for this phenomenon will be discussed shortly:

For potentials higher than 0.3 V (S.H.E.):
Aluminium reacts readily with water to form a hydrous oxide film. The essential steps of the aluminium + water reaction appear to be: (i) formation of amorphous oxide, (ii) dissolution of this oxide, and (iii) precipitation of the dissolved species as hydrous oxide.
In step (i) the amorphous oxide is continuously replenished by an electrochemical process in which barrier film growth is the anodic half-cell reaction coupled with the cathodic reduction of water (ALWITT [8]). This means that dissolution of aluminium occurs first, followed by hydrous oxide formation. Dissolution occurs at the crack tip, where bare material will be exposed to the electrolyte. The crack flanks will be protected by the precipitated oxide layer, which means less corroded and no tapered flanks (Figure 16).

For potentials lower than 0.3 V (S.H.E.):
The bad influence from cathodic potentials (till -1.8 V) on the fatigue life might be explained as follows:
(i) Dissolution from material at the crack tip. Crack surfaces formed at cathodic potentials show severe pitting corrosion, (Figure 17), which is missing at rest-potential conditions (Figure 18).
(ii) Diminishing inhibitor properties of the solution.
(iii) Increasing ad- or absorption of hydrogen ions.

4. ANALYSIS OF THE CRACK GROWTH MECHANISM
A model used first by KELLY [3] et al and later by WESTWOOD [9] et al was applied to liquid metal embrittlement. This model is extremely useful also to explain the corrosion fatigue mechanism.
The crack in Figure 19 will either propagate rapidly by cleavage or slowly by shear, depending on whether the tensile fracture stress \( \sigma_{cr} \) of the bond \( A - A_o \) constituting the crack tip, or the shear stress \( \tau_{cr} \) to cause
dislocation motion on the most favourably oriented slip plane $S - P$, is exceeded first. The tendency to fail by cleavage increases as the ratio $\frac{\sigma_{cr}}{\tau_{cr}}$ decreases. This can be achieved for example by cold work (increasing $\tau_{cr}$) or by a reduction of $\sigma_{cr}$. In general the environment will reduce $\sigma_{cr}$ and this is the basic of the corrosion fatigue mechanism as proposed in this chapter.

There are two important reasons for a reduction of $\sigma_{cr}$ by the environment:

1. Lowering of the surface energy by ion adsorption at the metal surface or by electrochemical reactions at the crack tip like metal dissolution and oxide film formation.

2. Diffusion penetration of hydrogen.

It is the competition between $\tau_{cr}$ and $\sigma_{cr}$ which determines the crack growth behaviour. The fracture mechanism will depend on the question whether $\tau$ or $\sigma$ is more critical.

In relation to this competition between $\tau_{cr}$ and $\sigma_{cr}$ it is significant to note on which planes $\tau$ and $\sigma$ will be maximal. For the tensile stress this will be a plane ($x/z$ plane) perpendicular to the loading direction and perpendicular to the plane of the specimen ($x/y$ plane). For the shear stress however, this plane is not perpendicular to the loading direction and even more important, it is not perpendicular to the plane of the specimen (SCHIJVE [10]). Since planes making an oblique angle to the plane of the specimen are carrying the maximum $\tau$ the tendency for growing in the $x/z$ plane will not be promoted if $\tau$ is controlling the mechanism. This is explaining the rougher fracture surface (multiple shear) in vacuum.

It also explains the larger flat area ($x/z$ plane) if $\sigma_{cr}$ is controlling the mechanism, which applies if the environment contributes to the crack extension.

Fatigue crack propagation is the result of localised plastic deformation. In combination with other conditions such as corrosive environment and low temperature the magnitude of $\sigma_{cr}$ may change to such an extent that the nature of the cracking process changes from a ductile shear mechanism to a cleavage or quasi-cleavage phenomenon.

In high vacuum any environmental effect is excluded and as a result $\tau_{cr}$
will determine the crack growth behaviour. The nature of the cracking process is fully ductile. This is in agreement with a number of observations: (i) A rough macroscopic fracture surface. (ii) Irregular, wavy striations. (iii) Dense networks of dislocations. Only for fatigue in air was it possible to relate substructure band spacing to cyclic crack growth (striations). However, in vacuum more difficult extension, involving more cross-slip, will lead to dense networks of dislocations (WANHILL [7]). (iv) Low crack propagation rates.

In more aggressive environment $\sigma_{cr}$ will play a more important role in the crack growth mechanism. The nature of the cracking process is brittle. The crack extension in one cycle consists of a zone of brittle fracture and a zone of ductile fracture. Ion adsorption, metal dissolution or oxide film formation will reduce the energy necessary for crack formation on a cube face and will therefore favour crack growth and striation formation on a plane suitably oriented to the direction of maximum tensile stress (STUBBINGTON [11]). Characteristic for this kind of crack growth are: (i) Macroscopically a flat tensile mode area. (ii) Well-defined parallel striations consisting of a flat region and a sharp trench. (iii) A well-defined substructure with periodic dislocation bands corresponding to the striation spacings (BROEK/BOWLES [12]), (iv) Significantly higher crack propagation rates compared with vacuum environment.

The large difference between crack growth rates in air and vacuum is easily understood using the above model. Oxidation of bare material at the crack tip will lower the surface energy and favours more brittle crack extension. It is the author's opinion that inefficient slip reversal because of the presence of an oxide layer as stated by many investigators is not the reason for this phenomenon.

The model of Tomkins appears to be useful to explain an environmental contribution to crack extension. However the Tomkins model fails to account for several observations mentioned before. For instance this model does not give any indication for the macroscopically rougher fracture surface (multiple shear mode) in vacuum. It does not explain why the environment should promote a tensile mode failure and why a shear mode failure will occur if $\tau_{cr}$ is controlling the fracture mechanism.
As a consequence of the proposed fatigue mechanism a striation consists of a region of brittle or quasi-brittle fracture (tensile component) and a region of ductile fracture (shear component). The more aggressive the environment the larger the brittle part of the striation.

5. CONCLUSIONS

1. A unique correlation between the state of stress and the mode of cracking does not exist, because the environment has a significant effect on the mode of cracking. The tensile mode area is larger the more aggressive the environment.

2. In inert environments crack extension is primarily dependent on plastic deformation. Nevertheless crack growth in vacuum can still occur cycle by cycle.

3. Competition between $\tau_{cr}$ and $\sigma_{cr}$ determines the fatigue crack growth behaviour.

4. Crack propagation rates in air are larger than in vacuum because of oxidation of bare material at the crack tip, which lowers the surface energy and favours more brittle crack extension.

6. ACKNOWLEDGEMENT

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<table>
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<tr>
<th>-sheet thickness (mm)</th>
<th>fatigue loading (kg/mm²)</th>
<th>wave form</th>
<th>frequency (Hz)</th>
<th>temperature (°C)</th>
<th>applied potential (volts versus SHE)</th>
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<tbody>
<tr>
<td>vacuum 2.10^-6 torr</td>
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<td>5 ± 2.5</td>
<td>s</td>
<td>0.5, 5</td>
<td>room</td>
</tr>
<tr>
<td>dry nitrogen &lt;170 ppm H₂O</td>
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<td>8 ± 4</td>
<td>s</td>
<td></td>
<td></td>
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<tr>
<td>dry oxygen &lt;200 ppm H₂O</td>
<td>2.5</td>
<td>8 ± 4</td>
<td>s</td>
<td>20</td>
<td>room</td>
</tr>
<tr>
<td>vacuum 5.10^-3 torr</td>
<td>2.5</td>
<td>8 ± 4</td>
<td>s</td>
<td>20</td>
<td>room</td>
</tr>
<tr>
<td>normal air</td>
<td>1, 2.5, 4</td>
<td>5 ± 2.5</td>
<td>s</td>
<td>0.5, 5</td>
<td>room</td>
</tr>
<tr>
<td>buffer solution</td>
<td>1.2</td>
<td>6 ± 4</td>
<td>s</td>
<td>5</td>
<td>room</td>
</tr>
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<td>buffer inhibitor solution</td>
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<td>6 ± 4</td>
<td>s</td>
<td>5</td>
<td>room</td>
</tr>
<tr>
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<td>6 ± 4</td>
<td>s</td>
<td>5</td>
<td>4, 20</td>
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<td>0.5, 5</td>
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<td></td>
<td>8 ± 4</td>
<td>s</td>
<td></td>
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</tbody>
</table>

**TABLE 1**

FATIGUE CRACK PROPAGATION TESTS
FIG. 1: TRANSITION FROM TENSILE MODE TO SHEAR MODE.
FIG. 2. FRACTURE SURFACE OF A CENTRE CRACKED SHEET SPECIMEN
FIG. 3: CRACK GROWTH RATE AS A FUNCTION OF $\Delta K$.
INFLUENCE OF THE SHEET THICKNESS.
FIG. 4. FRACTURE SURFACES OF FATIGUE SPECIMENS IN VACUUM, NORMAL AIR AND SALT WATER.
<table>
<thead>
<tr>
<th>Environment</th>
<th>Fatigue Life (k.c.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.5% NaCl</td>
<td>11</td>
</tr>
<tr>
<td>distilled water</td>
<td>21</td>
</tr>
<tr>
<td>buffer-inhibitor solution</td>
<td>26</td>
</tr>
<tr>
<td>normal air</td>
<td>27</td>
</tr>
<tr>
<td>vacuum 6.10^3 torr.</td>
<td>50</td>
</tr>
<tr>
<td>nitrogen oxygen</td>
<td>56</td>
</tr>
<tr>
<td>vacuum 2.10^6 torr.</td>
<td>60</td>
</tr>
</tbody>
</table>

**FIG. 5: FATIGUE LIFE IN DIFFERENT ENVIRONMENTS.**
FIG. 6: CRACK GROWTH RATE AS A FUNCTION OF $\Delta K$
FIG. 7. MACROSCOPIC FRACTURE SURFACES OF CENTRE CRACKED SHEET SPECIMENS IN VACUUM AND AIR RESPECTIVELY.
FIG. 8. MACROSCOPIC FRACTURE SURFACES OF FRETTING FATIGUE SPECIMENS IN VACUUM AND AIR RESPECTIVELY.
FIG. 9. CROSS SECTION OF A FRACTURE SURFACE IN VACUUM

FIG. 10. CROSS SECTION OF A FRACTURE SURFACE IN AIR
FIG. 11. FATIGUE STRIATIONS IN VACUUM AT A $\Delta K$ OF 32 KG/MM$^{3/2}$, $\times 10.500$
FIG. 12. FATIGUE STRIATIONS IN VACUUM AT A $\Delta K$ OF 46 KG/MM$^{3/2}$, $\times 13,000$
FIG. 13. FATIGUE STRIATIONS IN AIR, AT A $\Delta K$ OF 32 KG/MM$^{3/2}$, x 4000
FIG. 14: CRACK GROWTH RATE AS A FUNCTION OF $\Delta K$. 

7075-T6
$\sigma = 8 \pm 4 \text{kg/mm}^2$
$R = 1/3$
3.5% NaCl
FIG. 15: FATIGUE LIFE AT DIFFERENT APPLIED POTENTIALS.
FIG. 16. FATIGUE CRACK IN A BUFFER INHIBITOR SOLUTION AT AN APPLIED POTENTIAL OF 25 V (S.H.E.), × 1000.
FIG. 17. FRACTURE SURFACE FORMED AT AN APPLIED POTENTIAL OF -1.2 V (S.H.E.),
$\times 10000$. 
FIG. 18. FRACTURE SURFACE FORMED AT REST-POTENTIAL CONDITIONS, x 10,000.
Fig. 19: Schematic illustration of a crack in a solid subjected to an increasing force F. The bond A – Ao constitutes the crack tip. B is a surface liquid metal atom. (Westwood et al. 9).