Crack growth behaviour of aluminium wrought alloys in the Very High Cycle Fatigue regime

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Abstract. Investigations have shown that in the regime of Very High Cycle Fatigue (VHCF) “natural” crack initiation often takes place underneath the material surface leading to crack propagation without contact to atmospheric components. In order to elucidate the environmental damage contribution and its effect on the VHCF long crack propagation, fatigue experiments with alternating environment (vacuum and laboratory air) were performed. An ultrasonic fatigue testing system (USFT) equipped with a small vacuum chamber was applied that enables the in-situ examination of the long fatigue crack propagation at a resonance frequency of about 20 kHz by using a long distance microscope. By means of the Focused-Ion-Beam technique, micro-notches were prepared in the USFT specimens. The tests were carried out on the aluminium alloys EN-AW 6082 and 5083 in different conditions. It has been found that the atmosphere has a significant influence on the VHCF long crack propagation which manifests itself in the crack path as well as in the crack growth rates. Because of pronounced single sliding in vacuum, shear-stress-controlled crack propagation was detected whereas in laboratory air normal-stress-controlled crack propagation occurred. Furthermore, it has been proven that the secondary precipitation state of the aluminium alloy significantly influences the VHCF long crack propagation in vacuum.

1 Introduction

In the regime of Very High Cycle Fatigue (VHCF), that means at very low stress amplitudes, “natural” cracks often initiate internally underneath the material surface [1, 2]. Internal cracks propagate without contact to atmospheric components, thus, crack propagation in vacuum can be assumed. However, up to now it is unknown how the fatigue crack propagation in the VHCF regime exactly takes place and it is questionable, if all the crack growth phases of the conventional crack growth curve occur depending on this condition.

Fig. 1. Conventional fatigue crack growth curve [3].

Fig. 1 shows the conventional fatigue crack growth curve. In the short crack regime, the fatigue crack

propagation is highly influenced by the microstructure, therefore, the crack propagation rates reveal a strong fluctuation. In the long crack regime, however, the fatigue cracks propagate continuously and can be described by the Paris-Erdogan law.

It is a well-known fact that the humidity of ambient air deteriorates the fatigue properties of aluminium alloys compared to an inert environment [4-6]. In this study, the atmospheric influence on the VHCF long crack propagation in the aluminium wrought alloys EN-AW 6082 and 5083 in different material conditions is investigated. Fatigue experiments with changing environment between laboratory air and vacuum were performed by means of the ultrasonic fatigue testing technique equipped with a small vacuum chamber. The detrimental effect of water vapour in ambient air as well as the effect of alloy microstructure on the long crack propagation will be discussed in this article.

2 Materials and methods

In this study, the aluminium wrought alloys EN-AW 6082 and EN-AW 5083 were investigated whereby the precipitation hardenable alloy was examined in the peak-aged (pa) as well as overaged (oa) condition. The heat treatment was performed at the Technische Universität Dresden. The solid solution strengthened aluminium

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alloy EN-AW 5083 H111 has been strain-hardened and represents the precipitate-free condition.

Table 1. Ultimate tensile strength (R_m), yield stress ($R_{p0.2}$) and Vickers Hardness for the investigated aluminium alloys.

	R_m (MPa)	$R_{p0.2}$ (MPa)	Vickers Hardness
EN-AW 6082 (pa)	366	344	145 HV0.2
EN-AW 6082 (oa)	294	262	105 HV0.2
EN-AW 5083 H111	270	110	106 HV0.2

Figure 2 was generated at the Technische Universität Dresden by means of the transmission electron microscopy (TEM) showing secondary precipitates of the aluminium alloy EN-AW 6082 in peak-aged (Fig. 2a) and over-aged (Fig. 2b) condition. Due to the longer heat treatment, the secondary precipitates in the over-aged condition are coarsened.

Fig. 2: Secondary precipitates of the aluminium alloy EN-AW 6082: a) peak-aged and b) over-aged condition (generated at the Technische Universität Dresden by means of TEM).

The fatigue crack growth experiments were carried out using an ultrasonic fatigue testing system (USFT) equipped with small vacuum chamber (Fig. 3).

Fig. 3. In-situ setup of the USFT equipped with a small vacuum chamber.

The fatigue experiments were performed at resonance frequency of about 19.2 kHz under fully reversed tension-compression loading ($R = -1$). The pulse and pause sequences were adjusted for the fatigue in vacuum in order to prevent that the material temperature exceeds 30 °C. Fatigue experiments with changing environment (vacuum and laboratory air) were performed by means of interspersed testing sections in air. The ultrasonic specimens include a shallow-notch in order to monitor the fatigue crack propagation on the

material surface by using a long distance microscope. Prior to the fatigue experiments a micro-notch was milled into the mechanically and electrolytically polished ultrasonic specimens by means of the Focused-Ion-Beam (FIB) technique. In principle, the shape of the FIB micro-notch was chosen according to the research results of the Universität Kassel (Fig. 4a) [7].

Fig. 4. Shape of micro-notch: a) schematic illustration of a micro-notch used at the Universität Kassel [7] and b) FIB micro-notch prepared at the Universität Siegen.

Figure 4b shows a detail view of the depth difference at the end of the FIB micro-notch leading to high local stress concentration during cyclic loading. Hence, crack initiation can be successfully realized even at very low stress amplitudes at both ends of the FIB micro-notch.

3 Results

Fatigue experiments with changing environment were performed for all the aluminium wrought alloys. Therefore, four characteristic areas result in the crack path named as Air 1, Vacuum 1, Air 2 and Vacuum 2.

3.1 Crack propagation behaviour of EN-AW 6082 (pa)

A FIB micro-notch with a length of 120 µm was prepared in an ultrasonic specimen. In Fig. 5 the crack path, the EBSD analysis and the corresponding crack growth rates are shown. Crack initiation took place under constant stress amplitude of 125 MPa in laboratory air. In the beginning part in section Air 1, the right crack tip started to propagate at an angle of about 45° to the loading axis and shows a correspondence to the calculated glide trace with the highest Schmidfactor. But after reaching a certain crack length, crack propagation perpendicular to the loading axis was detected. After that, the environmental condition was changed to Vacuum 1. In Fig. 5a, slip bands (blue ellipse) are clearly formed on the material surface corresponding on the right side to the glide trace with highest Schmidfactor. Furthermore, the crack path on both sides deviates significantly from the normal direction. On the left as well as right side, it can be seen that the crack followed the glide trace with the second highest Schmidfactor. Moreover, a comparable behaviour is shown in a sub-grain (marked as red rectangle in Fig. 5b) at the left crack path. A correlation between the slip band formation on the material surface and the calculated glide trace with the highest Schmidfactor is reconfirmed.

Fig. 5. Crack path in EN-AW 6082 (pa) at $\Delta\sigma/2 = 125$ MPa and changing environmental conditions: a) slip band formation on the material surface, b) EBSD-analysis with calculated glide traces and c) crack growth rates of the left and right crack tip.

In section Air 2, the crack propagated perpendicularly to the loading axis and no correlation between the crack path and the calculated glide traces is observed. In section Vacuum 2, the crack path again shows significant differences to the crack propagation in air. Within the grain on the right side of the crack tip, strong crack deflection and slip band extension of the crack path is detected (Fig. 5a and b). Moreover, a clear correlation between the crack path (left and right) and the calculated glide traces exists. The crack propagation rates are shown in Figure 5c. In vacuum the crack growth rates are significant smaller than those in air. The differences between air and vacuum can be also recognised in the fracture surface morphology which is shown in Fig. 6.

Fig. 6. Fracture surface morphology: a) overview and b) detail views of the section Vacuum 2 with faceted structures.

In Fig. 6a, the sections Air 1 and Air 2 show fatigue fracture surfaces which are homogeneous and differ significantly from those observed in the sections

Vacuum 1 and Vacuum 2. In the detailed views in Fig. 6b in the section Vacuum 2, areas where the fracture surface morphology shows faceted structures are apparently visible.

3.2 Crack propagation behaviour of EN-AW 6082 (oa)

Prior to the fatigue experiment, a micro-notch with a length of 260 μm was introduced to an ultrasonic specimen of the aluminium wrought alloy EN-AW 6082 in the overaged condition. In Fig. 7, the crack path, the EBSD-analysis with the calculated glide traces for each grain and the crack growth rates for both crack tips are shown. A crack was initiated at constant stress amplitude of 120 MPa in laboratory air. In section Air 1, it can be seen that the crack path is pronounced perpendicular to the loading axis and does not show any correlation with the calculated glide traces. Due to the change of the environmental condition (Vacuum 1), slip bands (Fig. 7a blue ellipses) formed on the material surface corresponding clearly to the calculated glide traces of the individual grains. Moreover, on the left side of the crack path strong crack deflection was detected in some sections (Fig. 7a red arrows) whereby a correspondence between the crack path and the calculated glide trace with the highest Schmid factor can be observed. On the right side, however, the crack propagated perpendicularly through the complete section.

Fig. 7. Crack path in EN-AW 6082 (oa) at $\Delta\sigma/2 = 120$ MPa and changing environmental conditions: a) slip band formation on the material surface, b) EBSD-analysis with calculated glide traces and c) crack growth rates of the left and right crack tip.

In section Air 2, crack propagation perpendicular to the loading axis was detected again whereby the crack path does not show any correspondence with the calculated glide traces. In the final section Vacuum 2, the crack predominantly grew in the normal direction. However, slip band formation on the material surface (Fig. 7a white ellipses) took place only very occasionally and localized at the crack tip in specific sections. Here, a correlation with the calculated glide traces is observed. In Fig. 7c, the corresponding crack propagation rates are

shown. Also here it can be noted that the crack growth rates in air are significant higher than those in vacuum. In Fig. 8 the fracture surface morphology of the fatigue experiment in Fig. 7 is shown.

Fig. 8. Fracture surface morphology: a) overview and b) faceted structures in particular sections.

The fracture surface morphology in Air 1 as well as Air 2 is continuously smooth and homogeneous. Similar surface structures were also predominantly found in the area Vacuum 1 and Vacuum 2. However, there are sections where the fracture surface morphology deviates significantly and shows faceted structures (Fig. 8b).

3.3 Crack propagation behaviour of EN-AW 5083 H111

As for the aluminium alloy described above, a FIB micro-notch with length of 260 μm was introduced in the ultrasonic specimen of the aluminium wrought alloy EN-AW 5083 H111. The damage behaviour as a consequence of the material fatigue is shown in Fig. 9.

Fig. 9. Crack path in EN-AW 5083 H111 at $\Delta\sigma/2 = 120$ MPa and changing environmental conditions: a) slip band formation on the material surface with calculated glide traces, b) EBSD-analysis and c) crack growth rates of the left and right crack tip.

At the initial stage after crack initiation in laboratory air, the fatigue crack continuously propagated in normal direction through the whole crack path from the macroscopic perspective independent on the atmospheric condition (Fig. 9b). However, slip band formation took place in specific areas which clearly shows a correlation with the calculated glide traces (see Fig. 9a, section Vacuum1 and Vacuum 2). For instance, at the end of the left crack tip in Vacuum 2, slip band formation on the material surface corresponds clearly to the glide trace with the highest Schmidfactor. Furthermore, significant differences can be seen by comparing the crack propagation rates in air and vacuum. Obviously, the crack growth rates are higher in laboratory air than in vacuum by a factor of ten (Fig. 9c). Figure 10 shows the fracture surface of the crack path shown in Fig. 9.

Fig. 10. Fracture surface morphology: a) overview of areas depending on the environmental condition and b) specific section in Air 2 showing striations.

At first glance, the fracture surface does not show significant differences between air and vacuum. However, there are particular areas in air where striations were locally found (see Fig. 10b). In general, the fracture surface is continuous smooth, homogenous and pronouncedly perpendicular to the applied normal stress.

4 Discussion

4.1 Atmospheric influence

The results of the fatigue experiments with changing environment clearly demonstrate an atmospheric effect on the VHCF long crack propagation. Especially in the aluminium alloy EN-AW 6082 in the peak aged condition, the differences can be detected in the crack path, the fracture surface and the crack growth rates. Due to the change of the environment between air and vacuum, it becomes obviously that not only the crack propagation rates clearly decrease in vacuum, but also the crack propagation mechanism changes significantly. Thus, after reaching a certain crack length in air the

VHCF long crack propagation takes place consistently normal-stress-controlled. The incoherence between the calculated glide traces and the crack path in Fig. 5b confirms this behaviour. Furthermore, the preferred Stage II crack propagation in air also manifests itself in the fatigue fracture surface in Fig. 6 showing very smooth and homogeneous structures pronouncedly perpendicular to the loading axis. The answer to the question whether the VHCF long cracks can propagate consistently in laboratory air, depends significantly on the spatial distribution of the primary precipitates and the stress intensity in front of the crack tip [8]. In contrast, very pronounced single sliding occurs in vacuum due to the absence of air humidity. Based on the results from the fatigue experiment shown in Fig. 5a, it can be concluded that slip band formation on the material surface and shear-stress-controlled VHCF long crack propagation along an activated slip band take place even if the crack is very long. Further, a clear correlation between the calculated glide traces, the slip band formation on the material surface and the crack path is confirmed (Fig. 5b). Moreover, in contrast to the fatigue crack propagation in air, the preferred Stage I long crack growth in vacuum is observed on the fatigue fracture surface which shows crystallographic facets corresponding to the $\{111\}$ planes of aluminium. Therefore, it can be concluded that the damage mechanism during material fatigue in vacuum is significantly different from that in air. It is assumed that water vapour as a constituent of ambient air can rapidly reach the crack tip and adsorb on newly formed surfaces that form as a consequence of the crack progression resulting in a local chemical reaction. The formation of an aluminium oxide layer leads to water reduction at the interface between the oxide and the metal leading to the deposition of chemisorbed atomic hydrogen [5, 11]. A certain amount of the atomic hydrogen can diffuse through the plastically deformed zone in front of the crack tip during cyclic loading, if the crack propagation rate (below 10^{-9} m/cycle) is smaller than the mean diffusion distance of hydrogen at room temperature (3×10^{-9} m) [9, 10]. In this case, the hydrogen diffusion causes to an embrittlement effect [10, 11] at the crack tip which means that the atomic bonding can be locally reduced. The cyclic slip irreversibility will consequently be increased during cyclic loading so that multiple sliding is favoured in the plastically deformed zone. Thus, the VHCF long crack propagation takes place normal-stress-controlled without the slip band formation on the material surface. The fatigue properties of the investigated aluminium alloys deteriorate significantly due to the aforementioned mechanism and the resulting irreversible plastic deformation manifests itself in the crack growth rates. The values of the crack growth rates in air (above 10^{-10} m/cycle) correspond to the inter-atomic distance of aluminium (2.86×10^{-10} m) and hence, 1 Burgers Vector [8, 9]. In vacuum crack propagation rates between 10^{-12} and 10^{-11} m/cycle were detected. In other words, the reversibility of the plastic deformation in vacuum is much higher that explains the lower crack propagation rates below the inter-atomic distance of aluminium. As mentioned above, very pronounced single

sliding is promoted in vacuum because of the absence of water vapour. This assumption is confirmed due to the experimental observations showing slip band formation on the material surface near the crack tip during material fatigue in the VHCF regime. However, the results of the fatigue experiments clearly demonstrate that the damage behaviour in vacuum is strongly affected by the secondary precipitation state of the hardenable aluminium wrought alloy EN-AW 6082.

4.2 Influence of the precipitation state on the VHCF long crack propagation in vacuum

The different heat treatments of the aluminium wrought alloy EN-AW 6082 lead to precipitation of a secondary phase. The particles of the secondary phase differ essentially in their shape and size. In particular, the microstructure in the peak-aged condition exhibits semi-coherent fine cuttable secondary precipitates with a preferred orientation, while non-coherent larger particles are presented in the over-aged condition. It can be assumed that the principal glide mechanism of dislocations during material fatigue in vacuum is significantly influenced by the secondary precipitated particles which, in turn, determine the VHCF long crack propagation. In the case of the peak aged condition, the cutting mechanism of the semi-coherent particles promotes a to-and-fro glide of dislocations on specifically favoured glide planes whereby the activation of the glide planes depends on the individual grain orientation to the loading axis [12]. Therefore, very intensive local plastic deformation takes place within individual slip bands along the crack front due to very pronounced single sliding causing shear-stress-controlled VHCF long crack propagation in EN-AW 6082 (pa). The coarsening of the secondary precipitated particles in the over-aged condition caused by the Ostwald ripening leads to a significant change of the interaction between the dislocations and particles. It is assumed that in this configuration dislocations preferentially bypass the non-coherent particles due to energetic reasons so that cross slipping is expected to be the main glide mechanism. Thus, several slip systems are activated leading to a uniformly deformation at the crack tip and, hence, normal-stress-controlled VHCF long crack propagation in vacuum. However, few areas with shear-stress-controlled crack propagation were also found in the crack path as well as in the fatigue fracture surface (Fig. 7a and 8b) where the cutting mechanism of the secondary particles was most probable preferred.

Similar findings were observed by Richard et al. [5] who summarized the results of different studies on the fatigue long crack propagation in metallic alloys. They concluded that two intrinsic regimes (named as intrinsic stage I-like and intrinsic stage II regime) exist in high vacuum which differ significantly in respect to the crack growth mechanism. Thus, crack propagation in the intrinsic stage I-like regime preferentially takes place in microstructures which contain fine cuttable precipitates leading to heterogeneously localized deformation within slip bands in the near of the crack tip. As a result, faceted

fracture surfaces and strong crack deflection in the crack path occur. The intrinsic stage II crack propagation preferably takes place in microstructures containing semi-coherent or non-coherent precipitates. In this microstructure configuration, several slip systems are activated during cyclic loading resulting in a homogeneous deformation in front of the crack tip and a smooth crack path.

The microstructure of the aluminium alloy EN-AW 5083 H111 as the secondary precipitation free condition exhibits an increased dislocation density due to strain-hardening. As a consequence of the solid solution hardening, in general the aluminium alloy is expected to prevent wavy slip due to the decrease of the stacking-fault energy [12]. However, it is assumed that the high dislocation density in the material has a dominant effect on the plastic deformation in front of the crack tip leading to the activation of several slip systems during cyclic loading. Consequently, only normal-stress-controlled crack propagation takes place leading to a smooth crack path and a fatigue fracture surface pronouncedly perpendicular to the loading axis (Fig. 9b and 10).

5 Conclusion

Fatigue experiments with a changing environment between air and vacuum were performed on the aluminium wrought alloys EN-AW 6082 in the peak-aged and over-aged condition as well as EN-AW 5083 H111 by means of the ultrasonic fatigue testing technique. The main findings are summarized as follows:

- The fatigue experiments with a changing environment clearly show that the atmospheric conditions have a strong effect on the material damage. In vacuum, very pronounced single sliding takes place leading to shear-stress-controlled VHCF long crack propagation in the aluminium alloy EN-AW 6082 in the peak-aged condition. The VHCF long crack propagation in vacuum in the peak-aged condition differs, therefore, significantly from the conventional long crack growth in the Low Cycle or High Cycle Fatigue regime.
- The stage I crack propagation in EN-AW 6082 (pa) in vacuum is independent on anisotropic effects as rolling texture and takes place due to the absence of the water vapour in ambient air.
- It is assumed that water vapour in ambient air can rapidly reach newly formed crack surfaces during material fatigue leading to an embrittlement effect in the plastically deformed area, if the crack propagation rate is smaller than the mean diffusion distance of hydrogen. As a result, normal-stress-controlled VHCF long crack propagation (Stage II) takes place after reaching a certain crack length in laboratory air.
- The larger irreversibility of the plastic deformation in laboratory air compared to that in vacuum manifests

itself in the crack propagation rates. Thus, crack propagation rates in the range of 10^{-10} m/cycle were detected in air corresponding to the inter-atomic distance of aluminium. The value of the crack growth rates in vacuum (10^{-12} - 10^{-11} m/cycle) are related to the larger reversibility of the plastic deformation in front of the crack tip.

- The results clearly demonstrate that the VHCF long crack propagation in vacuum is strongly influenced by the secondary precipitations in EN-AW 6082. Thus, very intensive and localized plastic deformation within slip bands takes place in the peak aged condition leading to stage I crack propagation. In this case, the cutting mechanism of the semi-coherent particles enables a to-and-fro glide of dislocations on specifically favoured glide planes. Due to the coarsening of the secondary particles in the over-aged condition, dislocations preferentially bypass the non-coherent precipitates. This leads to a homogeneous plastic deformation along the crack front and, hence, normal-stress-controlled crack propagation. However, shear-stress-controlled VHCF long crack propagation in the over-aged condition can also occur in specific areas, if the secondary particles are locally cuttable.
- In the solid solution and strain-hardened aluminium alloy EN-AW 5083 H111 basically multiple slipping takes place in the plastically deformed area in front of the crack tip supported by the increased dislocation density in the material. As a result, only stage II crack propagation was detected in this material condition.

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