Understanding the near-threshold crack growth behavior in an aluminum alloy by x-ray tomography

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Abstract. The near-threshold behavior of long cracks is evaluated using pre-cracked flat dogbone specimens of a commercial aluminum alloy in two heat treatment states. Once the threshold was known, the crack initially introduced by cyclic compression was propagated under load control with approximately constant range of the stress intensity factor at values close to the threshold values. The 3D morphology of the crack and the spatial distribution of the primary precipitates obtained by μ -CT were reconstructed with the aim of analyzing the role of the microstructure and primary precipitates on crack growth. Results pointed out that there were two major mechanism keeping the crack from continuous extension. First, the crack front was pinned by primary precipitates leading to kinking of the crack front. The second mechanism was pronounced shear-controlled crack growth of long cracks seen in terms of hilly structures on the crack faces. The first mechanism was found to be dominant at $R=-1$, whereas tests performed with $R=0.1$ showed an increased propensity towards shear controlled crack growth.

1 Introduction

Unusual crack extension mechanisms may occur if components containing pre-existing flaws are subjected to cyclic loads at very small amplitudes. This is related to the fact that the energy available for cyclic crack extension is very low in this case, and that the cyclic plastic zone ahead of the crack tip may be somewhat incomplete. Understanding the phenomenon implies that crack extension has to be studied in the near-threshold regime in a very well-controlled way.

The method proposed in ASTM E 647 standardizing the determination of the long-crack threshold is based on the so-called load-shedding procedure. However, it has been reported (e.g. [1-5]) that crack closure effects originating from the stepwise decrease of the stress intensity range can shift the threshold towards higher values causing non-conservative estimates. With the aim of circumventing the problems related with ASTM E 647, Pippan et al. [6, 7], among others, proposed an alternative procedure in which the threshold is approached from below. After pre-cracking specimens containing a sharp notch by cyclic compression, the load amplitude at the desired stress ratio R is increased in steps until the crack propagates stably. Cracks generated by compression pre-cracking and propagating at a constant stress amplitude have been shown to be well suited for analyzing crack extension at very low cyclic loads.

In a previous study [8] on the near-threshold behavior of long cracks in a commercial aluminum alloy at $R = -1$, it was found that local pinning of the crack

front caused by primary precipitates and shear-dominated crack extension very similar to the stage-I extension of small cracks keep the crack from extending continuously. The last effect was most surprising as the cracks analyzed in [8] were definitely long cracks, which extended over several grains. However, shear-dominated effects may have been enhanced since the tests in [8] were performed at $R=-1$ and hence not in a typical fracture mechanics testing scenario. Some minor crack closure effects may even have developed in the compression phase of the loading cycle when the crack faces were in contact making it worthwhile to verify the unexpected crack extension behavior found in [8] under tension-tension loading.

Therefore, fatigue crack growth tests at a stress ratio of $R = 0.1$ were performed in the present study using the aluminum alloy investigated in [8]. Two heat treatment states were examined with the aim of getting in-depth information on the influence of the microstructure and precipitates on crack growth.

The paper is structured as follows: The description of the material and the experimental methods in section 2 is followed by an overview of the fracture mechanics tests. In section 4, crack growth of long cracks in the near threshold region and crack advance of long cracks are discussed with some concluding remarks given in section 5.

2. Material and experimental methods

The material examined in this study is the aluminum alloy EN-AW 6082, which was peak-aged (6PA) and overaged (6OA). $R_{p0.2}$ and R_m were determined in a

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tensile test as 336 MPa and 344 MPa for material 6PA, and 262 MPa and 294 MPa for material 6OA, respectively. A rolled sheet with a thickness of 20 mm and the microstructure shown in Fig. 1 taken in rolling direction served as base material. The grains are elongated with a maximum extension of up to 2000 – 3000 μm in rolling direction and 100 - 200 μm in transverse direction. Two types of primary precipitates can be detected in the microstructure: Mg-based ones (dark spots in Fig. 1) and Fe-based ones (bright spots in Fig. 1), which are arranged in a line-like fashion in rolling direction.

Fig. 1. EN-AW 6082 SEM micrograph of grain structure taken in rolling direction

Flat dogbone specimens with a total length of 40 mm and the geometry shown in Fig. 2a were machined out of the sheet material both in rolling (-LS) and transverse direction (-TS). After mechanical and electrolytic polishing, a part-through notch with a depth of 50 - 150 μm depending on test conditions and a notch radius smaller than 20 μm was cut in the specimen radius using a razor blade polishing technique similar to that proposed by Nishida et al. [9], see Fig. 2b. This technique is common practice in ceramics, but can also be used for metals, if the load applied to the razor blade is small enough. The main advantage of this technique is that extremely sharp notches can be made exhibiting only minor plastic deformation.

Fig. 2. a) Specimen geometry and b) Part-through notch located at the specimen radius

Fatigue crack growth tests were performed at a stress ratio of $R = 0.1$ on a Rumul Mikrottron resonance machine equipped with a 20 kN load cell and a sinusoidal force with a frequency of ca. 150 Hz in laboratory atmosphere. The test procedure was quite similar to the one used in [8] and is described again for the sake of completeness.

A long-distance microscope was used to monitor the crack growth on the specimen surface, enabling the documentation of the crack propagation when the cyclic load is interrupted by stopping the testing machine. The crack growth rate was calculated using the measured crack length and the number of test cycles between the scans. The specimens were pre-fatigued in compression to introduce a pre-crack which is open when unloaded. The crack growth test was started after applying up to 500.000 cycles at a stress ratio of $R = 20$ and a stress level of $\sigma_{min} = -290 \text{ MPa}$ for material 6PA and $\sigma_{min} = -208 \text{ MPa}$ for material 6OA, respectively.

The threshold of the stress intensity range ΔK was determined using the stepwise increasing load amplitude crack growth test described in [7]. After pre-cracking in compression, testing was changed to a pull-pull load ($R = 0.1$) at a low stress intensity range of about $\Delta K = 0.6 \text{ MPa}\sqrt{\text{m}}$ in order to let the crack propagate. If no crack growth was detected after 500.000 cycles, the stress amplitude was increased by 5 - 10 %. This procedure was repeated until the crack propagated in a stable manner. The threshold was then defined using the stress amplitude which first led to continuous crack growth.

Once the threshold was known, specific values of the stress amplitude were selected corresponding to ranges to the stress intensity factor (SIF) close to the threshold value for the starting crack length of 250 μm . The crack was then propagated at a constant stress amplitude according to the procedure proposed in [8] until the crack advance, visible on the surface, amounted to 50 μm . The stress amplitude was then decreased such that the initial value of the range of the SIF was recovered.

This procedure, allowing a crack propagation at a nearly constant stress intensity range with a maximum variability of ΔK of 10 %, was repeated until the crack reached a length of 1000 μm or the crack growth rate fell below 10^{-11} m/cycle , which was defined as the criterion for a crack stop. The fractured specimens were analyzed both in a scanning electron microscope (Zeiss Ultra 55) and in high-resolution X-ray computed tomography ($\mu\text{-CT}$). The three-dimensional measurements were carried out with a ZEISS Xradia 520 Versa X-ray microscope at a voltage of 40 kV and a power of 3 W. Up to 2501 radiographs were captured with a resolution of about 3.4 $\mu\text{m/voxel}$ and the 4x objective. Application of image correction filters was followed by reconstruction of the acquired radiographs with TXM Reconstructor (Zeiss) software. The image analysis was carried out with the rendering software Avizo (VSG), in which the crack morphology and spatial distribution of the primary precipitates was segmented manually after applying the non-local means filter.

3. Experimental results

3.1 Long-crack threshold

Three tests on specimens of both materials in LS- and TS-orientation were performed to determine the long-crack threshold according to the procedures described in section 2. The resulting da/dN vs. ΔK -curve of one

experiment is shown in Fig. 3, where the cutting direction corresponds to the rolling direction (6PA-LS). Left to the dashed line representing the threshold value of ΔK_{th} , the crack stopped repeatedly as schematically illustrated by the circles on the x-axis, and the stress amplitude had to be increased by 5 – 10 %. After reaching the threshold value, the crack grew in a stable manner at a constant stress amplitude and showed long crack behavior.

Fig. 3. da/dN vs. ΔK -curve with threshold determination (ΔK_{th})

An overview of the determined threshold values, which are discussed in detail in [10], is given in table 1.

Table 1. Threshold data for material 6PA and 6OA.

	6PA-LS	6PA-TS	6OA-LS	6OA-TS
$\Delta K_{th,1}$	1.34	1.30	1.10	1.13
$\Delta K_{th,2}$	1.13	1.51	1.17	1.33
$\Delta K_{th,3}$	1.38	1.32	1.20	1.14
$\bar{\Delta K}_{th}$	1.28	1.38	1.16	1.20

Peak-aged material has been found to exhibit a higher threshold compared to the overaged material, which is in agreement with results presented in the literature for various aluminum alloys [11-13]. Specimens taken in transverse direction (-TS) have a slightly better near-threshold crack growth resistance as the samples machined in rolling direction (-LS) independent of the heat treatment condition.

3.2 Crack growth at $\Delta K = \text{const.}$

The influence of the microstructure and precipitates on crack growth in dependency of the heat treatment condition and specimen orientation was investigated in experiments at quasi-constant ΔK . On the basis of the determined thresholds, SIF values between 1.2 and 1.5 $\text{MPa}\sqrt{\text{m}}$ were defined.

3.2.1 Crack paths

A comparison of the crack paths of the investigated materials 6PA and 6OA tested at a ΔK -value of 1.3 $\text{MPa}\sqrt{\text{m}}$ is shown in Fig. 3. No results are depicted for material 6OA taken in rolling direction, as the crack repeatedly stopped at this load before reaching the starting crack length of 250 μm . The crack in the TS-oriented specimen of material 6PA stopped after ca. 310 μm , while it grew to a length of 410 μm (6PA) and 435 μm (6OA), respectively, in the LS-oriented samples. Crack arrest before reaching a crack length of 1000 μm was observed even in the experiments at higher load amplitudes.

No significant differences can be found between the crack paths of material 6PA in both LS- and TS-orientations and material 6OA in LS-orientation. The predominant crack propagation direction is mode-I with a certain amount of crack kinking (black arrows) or even crack branching when the crack path encounters primary precipitates. It can be seen that these precipitates are aligned in rolling direction (-LS) and form irregular clusters in transverse direction (-TS).

Taking a closer look at the crack paths reveals that there are occasionally sections in which the crack seems to propagate in a shear-dominated mode (white arrow in Fig. 4a). This effect is independent of the microstructure and rolling texture, as it was observed in both the LS- and TS-oriented specimens.

3.2.2 3D crack morphology

The 3D morphology of the cracks and spatial distribution of the primary precipitates obtained by $\mu\text{-CT}$ was reconstructed to further investigate the crack growth behavior at $\Delta K = \text{const.}$ Fig. 5 shows the results for the LS-oriented specimen of material 6OA tested at a load of 1.2 $\text{MPa}\sqrt{\text{m}}$. As it can be seen, the crack already stopped after ca. 400 μm on the observed specimen front shown on the left side of Fig. 5, while it grew to a length of about 1000 μm on the specimen backside

The reconstructed crack front is of irregular form exhibiting some kinks, which are related to discontinuities in the crack morphology (yellow ellipses in Fig. 5). A closer inspection reveals that these discontinuities are not restricted to kinks but showed up as isolated features on the crack faces too. In the near-surface region, the crack seems to propagate in a shear-dominated mode as indicated by the “hilly” features highlighted by the blue square.

3.2.3 Fracture surfaces

The fracture surfaces were analyzed to gain additional information on the crack propagation behavior in dependency of the heat treatment condition and specimen orientation. For this purpose, the specimens were cooled down in liquid nitrogen and cracked in tension in order to

Fig. 4. SEM-micrograph showing the crack paths at 1.3 MPa√m: a) 6PA-LS; b) 6PA-TS; c) 6OA-LS

Fig. 5. 3D crack morphology of material 6OA tested in LS-orientation at 1.2 MPa√m obtained by μ-CT

introduce as little plastic deformation as possible. In Fig. 6-9, the fracture surfaces of the samples in peak-aged and overaged state tested at a $1.5 \text{ MPa}\sqrt{\text{m}}$ are presented.

Fig. 6. SEM-micrograph showing the fracture surface of material 6PA-LS at 1.5 MPa√m

The fracture surface of material 6PA in LS-orientation is shown in Fig. 6. Only small irregularities (marked with orange arrows) can be detected along the crack front, which are related to fine white lines extending into the fracture surface. The final crack length on the observation side and specimen backside differs by about 40 μm. All in all, the fracture surface is smooth.

Fig. 7. SEM-micrograph showing the fracture surface of material 6PA-TS at 1.5 MPa√m

Different characteristics were detected on the fracture surface of the specimen taken from material 6PA in TS-orientation. As it can be seen in Fig. 7, a pronounced ridge parallel to the crack front is visible (blue square in Fig. 7) indicating that the crack extension is shear-dominated across the complete specimen. The difference in crack length between the observed specimen front and backside is about 180 μm.

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Fig. 8. SEM-micrograph showing the fracture surface of material 6OA-LS at 1.5 MPa√m

Fig. 8 shows the fracture surface of the LS-oriented specimen of material 6OA. While the crack grew to a length of ca. 435 μm on the observed specimen front, shown on the right side of Fig. 8, it already stopped after a length of ca. 120 μm on the specimen backside. Similar to specimen 6PA-LS, only small irregularities in the form of kinks extending as fine lines into the fracture surface can be detected.

Fig. 9. SEM-micrograph showing the fracture surface of material 6OA-TS at 1.5 MPa√m

The fracture surface of material 6OA in TS-orientation depicted in Fig. 9 shows significantly different characteristics. As it can be seen, the crack front lagged

Fig. 10. Close-up of the 3D crack morphology of specimen 6OA-LS tested at 1.2 MPa√m showing primary precipitates located at the root of ridges

behind on the specimen backside causing a difference in crack length of about 310 μm between the observed specimen front (right side in Fig. 9) and its backside. The distinct kinks in the irregular-shaped crack front are related to both lines extending into the fracture surface (orange arrows in Fig. 9) and ridges (green ellipses in Fig. 9). These ridges were not only restricted to kinks in the crack front, but showed up as isolated features on the fracture surface. A certain amount of shear-dominated crack growth can be observed in the near-surface region (blue square in Fig. 9).

4. Discussion

From the crack paths shown in Fig. 4, it can be concluded that the primary precipitates act as microstructural barriers causing deviations of the crack. These barriers apparently pinned the crack front at several locations as indicated by the 3D crack morphology (Fig. 5) and fracture surfaces (Fig. 6-9). A

closer inspection of the discontinuities in the 3D crack morphology of specimen 6OA-LS tested at $\Delta K = 1.2 \text{ MPa}\sqrt{\text{m}}$ reveals - after reconstruction of the 3D morphology of the primary precipitates - that large precipitates are located at the root of the ridges (see Fig. 10). These precipitates cause a pinning of the crack front eventually leading to a crack arrest.

Fig. 11. Close-up of the fracture surface of specimen 60A-TS tested at $1.5 \text{ MPa}\sqrt{\text{m}}$ showing lines of ductile fracture

Detailed analysis (see Fig. 11) of the white lines depicted in Fig. 9 indicates that these lines are ridges of ductile fracture decorated by dimples. The crack front was thus locally pinned and the crack faces were bridged by material ridges, which failed by ductile fracture.

The primary precipitates seem to have a stronger influence in the TS-direction as the crack front is held up several times. This effect may be correlated with the spatial distribution of the primary precipitates. As it can be seen in Fig. 12, the crack front passes through well-defined lines of primary precipitates decorating the grain boundaries in the LS-specimen, while local precipitate clusters are formed in the TS-oriented samples acting as pinning sites. Local propagation of the crack front is thus possible in precipitate free areas leading to kinked crack fronts (see Fig. 9). Overaging seems to increase the pinning potential of the primary precipitates as ridges of ductile fracture and pronounced crack front kinking were observed even in the LS-oriented specimen (see Fig. 5).

Fig. 12. CT scan and micrographs illustrating the position of the crack front according to the primary precipitates.

Even though a local pinning of the crack front was observed at $R = -1$, there are pronounced differences as to the pinning potential of the precipitates which occur in a less pronounced form at $R = 0.1$. This is attributed to the increased propensity of the primary precipitates to cracking rather than to pinning, as the crack front advance at $R = 0.1$ requires higher $K_{\max}$ -values at the same ΔK -value than a test at $R = -1$. Higher $K_{\max}$ -values also seem to enhance crack propagation in a shear-dominated mode according to the results presented above.

5. Conclusion

Results of near-threshold fatigue crack growth experiments at $R = 0.1$ on peak-aged and overaged aluminum specimens pre-cracked by cyclic compression were presented. Based on the long-crack thresholds initially determined on samples taken in both rolling and transverse direction using the load-increasing procedure proposed in [7], crack growth experiments were performed with a nearly constant ΔK -value resulting in a constant propagation rate. Detailed analysis of the

fractured samples in SEM and μ -CT points out that crack advance in a shear-dominated mode is enhanced at $R = 0.1$ compared to $R = -1$ [8] whereas the pinning potential of the primary precipitates is significantly decreased. This is related to the fact that the primary precipitates tend to crack with increasing R -ratio and are consequently less able to form substantial barriers for crack advance.

The authors would like to thank the German Research Foundation (DFG) for financial support.

6. References

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