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Microstructure investigations on two different aluminum wrought alloys after very high cycle fatigue

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ABSTRACT

Fatigue studies were conducted under load control with a servo-hydraulic testing machine with smooth specimens of the aluminum wrought alloys EN AW-6056-T6 (R_m = 399 MPa) and EN AW-6082-T5 (R_m = 356 MPa) up to a maximum number of $N = 2 \times 10^8$ and 10^9 cycles, respectively. The results show for both aluminum alloys that the fatigue strength decreases with increasing number of cycles after the kneepoint of the *S*–*N*-curve and that approximately at this point a transition of the crack initiation site from the surface to the subsurface occurs. All fractured specimens were investigated with a scanning electron microscope (SEM). Large defects like primary intermetallic particles could not be found at the crack initiation sites. The difference between the subsurface non-defect crack initiation sites of both alloys is to be found in the fractographic structure. The internal crack initiation site of EN AW-6056-T6 shows a multiplicity of cleavage-like planes contrary to the flat area found at the alloy EN AW-6082-T5. These results were already presented in [1]. The scope of this present paper is new microstructural investigations and the formulation of a failure model which respects the results of different scales of microscopy. Electron back scatter diffraction analysis (EBSD analysis) and transmission electron microscopy (TEM) were done for both alloys to characterize the microstructure and to compare unstressed and tested materials. These differences are discussed with regard to the differences in failure mechanisms.

1. Introduction

It is well known that the fatigue strength of materials with face-centered cubic lattice, as for instance aluminum, decreases with increasing number of cycles. Only a few sources are found in the public literature describing the cyclic strength of technical aluminum alloys [e.g. 2–4] in the VHCF-region and only few of them include investigations of the failure mechanisms [5–7]. The crack initiation site for aluminum wrought alloys was detected at the surface of the specimens [3,4,7] and close below the surface [2,5,6] in the VHCF-region. However, there are internal crack initiations which are not related to defects. Here the fracture mechanisms are more influenced by microstructure, interface and micro plasticity [3].

Within the scope of this paper, two different wrought aluminum alloys (EN AW-6056-T6, EN AW-6082-T5) were investigated to determine fatigue strength and failure mechanism at number of cycles up to $N = 2 \times 10^8$ and 10^9, respectively.

2. Methods

The investigations at EN AW-6056-T6 were done at smooth, mechanically polished specimens [1,8] at a stress ratio of $R = 0.1$. The smooth specimens of EN AW-6082-T5 were only precision-turned. They were investigated at $R = 0$ in previous studies [9]. At that time it was of interest to investigate the fatigue behavior with a technical surface condition. Fatigue tests were conducted under load control with a servo-hydraulic testing machine at a frequency of 400 Hz. The temperature of the specimens of EN AW-6056-T6 were checked on the first tested specimens in the very high cycles region (f = 400 Hz) about 10 times and also at two specimens fatigued at higher load (f = 50 Hz) for about 5 times with a thermometer with surface device. A linear regression was carried out to achieve a double σ–N line with

$$N = N_k \cdot (\sigma_{na}/\sigma_{Nk})^{-k} \tag{1}$$

and

$$N = N_k \cdot (\sigma_{na}/\sigma_{Nk})^{-k*} \tag{2}$$

Eq. (1) describes the σ–N-line for cycle numbers smaller than the kneepoint N_k and (2) the σ–N curve in in the VHCF-region. It

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is k for the slope, σ_{na} for the nominal stress amplitude and σ_{Nk} for the stress amplitude at the kneepoint N_k. The scatter for stress

$$T_\sigma = 1 : [\sigma_a(P_s = 10\%)/\sigma_a(P_s = 90\%)] = 1 : 1.25 \quad (3)$$

was taken from a recommendation in [10], with P_s as the probability of survival. The fracture surfaces of all specimens were investigated with SEM. The EBSD-analysis was used to measure the crystallographic orientation, to illustrate the crystallographic texture, to calculate the crystallographic orientation of the crack in the alloy EN AW-6056-T6 and to get information about the grain size and the schmid factor at the crack initiation site in the alloy EN AW-6082-T5. The precipitated hardening conditions and the grain boundaries were investigated by TEM.

3. Results

3.1. Material characterization

The main significant difference regarding the chemical composition of both alloys is that EN AW-6056-T6 is copper alloyed, Table 1, while AW-6082-T5 shows only traces of copper. It shows – despite of the larger grain size – the higher tensile strength and the higher ductility compared to EN AW-6082-T5, Table 2, Fig. 1. The specimens of both alloys were taken from a wire, so the grains are elongated in one direction. For determining the texture, it is necessary to know which crystallographic direction $[u\,v\,w]$ is parallel to the direction of pull which was arranged parallel to the tilt axis X0 of the SEM. So the texture is given by an inverse pole figure. The texture of EN AW-6082-T5 is a typical $\langle 1\,1\,1\rangle$ $\langle 1\,0\,0\rangle$ II X0 double fiber texture. The content of the $\langle 1\,0\,0\rangle$ oriented grains is about 50%, this [11] and the small mis-orientation angles inside the grains are indicating that the $\langle 1\,0\,0\rangle$ oriented areas are originated during recrystallization. EN AW-6056-T6 shows with a near $\langle 1\,1\,0\rangle$ II X0 texture a total different texture than EN AW-6082-T5. The alloy EN AW-6056-T6 has Q'-precipitates (Cu-alloyed) and precipitate free zones at the grain boundaries, Fig. 2a and b, while the alloy EN AW-6082-T5 does not show these features, Fig. 2c. In EN AW-6056-T6 exist needlelike precipitates which were oriented along $\{1\,0\,0\}$-planes, Fig. 3. It is known from [12] that the 6xxx alloy series have also precipitates at the $\{1\,1\,0\}$ planes, as can be seen on EN AW-6082-T5, Fig. 4. The strikes at $\{1\,0\,0\}$ planes – marked in Fig. 3b and c with arrows - can be interpreted as reflexes from the (semi) coherent lattice of the precipitates [13], indicating the presence of GPII-zones. The precipitates in EN AW-6082-T5 show no reflexes, i.e. no own lattice in the diffraction pattern. Those precipitates are GPI-zones. Therewith the differences in the tensile strength of both alloys can be explained. The maximum precipitation hardened condition and thus the higher tensile strength of EN AW-6056-T6 occur when GPII-zones exist. GPI-zones (EN AW-6082-T5) are typical for the underaged condition and cause a lower tensile strength. The main differences of both alloys are the grain size, the texture and the condition of the precipitates.

3.2. Fatigue strength and fracture surface analysis

The decrease of fatigue strength in the VHCF-region is similar for both aluminum alloys, Fig. 5. EN AW-6056-T6 in the peak-aged condition (GPII-zones) exhibits lower fatigue strength at $R \approx 0$ – despite a higher tensile strength – compared to EN AW-6082-T5 (GPI-zones). Failure starts at the surface at higher stress amplitudes and initiated internally as a so called non-defect failure [3] at lower stress amplitudes. The fractographic microstructures at the crack initiation sites of these two aluminum alloys are completely different. For EN AW-6056-T6, the crack initiation area as well as the whole fatigue fracture area show cleavage-like planes, Fig. 6. In fractography, the term cleavage is defined as a separation of crystallographic planes under normal tensile stress. In the present study, cleavage planes are not correlated to planes under maximum principal tensile stress. Therefore, the term cleavage-like plane is used. In the crack origin area but also in the further crack path triple junctions of grain boundaries were found. Thus, these triple junctions are not necessarily significant for the crack initiation. For EN AW-6082-T5 most of the internal crack initiation shows one flat area, Fig. 7a.

Table 1
Chemical composition (wt%).

	Si	Fe	Cu	Mn	Mg	Zn	Ti
EN AW-6056-T6	1.01	0.18	0.57	0.55	0.86	0.18	0.02
EN AW-6082-T5	1.13	0.25	0.02	0.73	1.1	0.02	0.01

Table 2
Mechanical properties.

	YS (MPa)	UTS (MPa)	A (%)	Z (%)	HV0.3
EN AW-6056-T6	344	393	19	47	131
EN AW-6082-T5	341	356	11	48	110

3.3. Microstructure in the crack initiation site

To investigate the microstructure in the internal crack initiation site and thus the failure mechanism, both alloys have been micro-sected through the crack initiation area and an EBSD-analysis has been conducted. For the micro sectioning the distance from the crack origin to the surface of the specimen were measured and documented in the SEM. With the SEM pictures the crack origin can be located exactly also under a stereo microscope. There the intersection line was marked at the surface of the specimen. The thickness of the specimen was measured with a micrometer gauge perpendicular to the mark. Then the specimen were sectioned with the cutting-machine Brilliant 221 from ATM in a distance to the crack origin of about 200 μm. The sectioned specimen was embedded in conductive epoxy resin with carbon. Measuring again results the distance which has to grinded and polished. Grinding was done with graining 120–180 and 220–320 with a force of 20 N and a time of 2 min. After this the specimen were measured again. So the procedure was iterative, measuring the specimen and grinding. Subsequently the specimen was polished with 9 μm, 3 μm and 0.2 μm.

For the alloy EN AW-6056-T6 the cracks have been found trans- and inter-granular, Fig. 6. The straight line in the two-dimensional micro-section, position 2 in Fig. 6c, is equivalent to the cleavage-like planes at the three-dimensional fracture surface. A second micro-section, Fig. 8, through the crack initiation area was done to verify the working hypothesis that the transgranular cleavage-like planes are oriented in special crystallographic planes which are perpendicular to each other, e.g. $\{1\,0\,0\}$ and $\{1\,1\,0\}$. The angle between the crack (cleavage-like plane) at positions 1 and 2 is 90°, so the crack changes the plane inside the grain. The secondary crack at position 3 is also a straight line and within a cleavage-like plane. Relevant for the calculation are the angles α_{mes} between the cleavage-like planes and the direction of pull (loading direction). Those data have been obtained from Fig. 8b and reported in Table 3. By EBSD-analysis, the Eulerian-angles Φ_1, ψ, Φ_2 were determined at the three positions marked in Fig. 8c and Table 3. From those measurements, the crystallographic direction $[u\,v\,w]$, which is defined parallel to the direction of pull [14] was calculated.

It can be shown that the measured angles between the cleavage-like plane and the direction of pull match the calculated angles