High-frequency cyclic testing of welded aluminium alloy joints in the region of very high cycle fatigue (VHCF)

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1. Introduction

The ultrasonic fatigue testing system (UFTS) technique with working frequencies from 18 to 21 kHz is nowadays a well-established method for cyclic material testing up to very high load cycles ($N_f > 10^7$). Latest research activities applying this system were so far focused on the damage mechanisms in macroscopically homogeneous material [1–3]. Discussion about the influence of microstructural inhomogeneities on the VHCF behaviour is mainly related to a microstructural scale (e.g. primary or secondary precipitations, non-metallic inclusions, second phases, pores in cast alloys, etc.). In this respect, the significance of crack initiation processes in quasi defect-free materials versus crack growth behaviour in defect afflicted materials for the reliability of fatigue life assessments are widely discussed. However, a study of complex structures with macroscopic defects deriving from manufacturing processes were not yet part of the ongoing research activities and will therefore be discussed in this paper using the example of welded aluminium alloy joints. The cyclic testing of such geometrically complex samples is still a challenge due to the specimen size and geometry restrictions of the UFTS [4,5]. In the case of analysing the VHCF behaviour of welded structures, both microscopic as well as macroscopic notch effects, such as the geometrical notch at the transition from base material to weld seam (weld reinforcement), the inhomogeneous microstructure with a mismatch of the monotonic strength over the different weld zones resulting from the process-related temperature gradient as well as typical weld seam imperfections (e.g. pores, incomplete fusions, hot cracks, etc.) must be taken into consideration.

Assessment of fatigue strength of welded structures has been extensively studied in the past and directives such as the recommendations of the International Institute of Welding (IIW) [6] and the Forschungskuratorium Maschinenbau (FKM) e.V. are well established [7]. Global and local approaches based on nominal, structural or notch stress methods as well as fracture mechanic models serve as origin for a stress analysis. For the design of aluminium joints, the local stress concept was successfully applied for welded structures of various Al-alloy combinations by Morgenstern [8] using a fictitious notch radius of $r_f = 1.0$ mm. The concept bases on experimental data for low cycle fatigue (LCF) and high cycle fatigue (HCF) region and is approximately assumed for VHCF range ($N_f > 10^7$). This approach for lifetime calculation is based on a master S–N curve for aluminium sheets with thicknesses between 5 and 25 mm irrespective of underlying strengthening mechanisms for the different Al-alloys.

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Störzel et al. [9] confirmed the concept of Morgenstern by comparing the reliability of global as well as local approaches to predict fatigue life of welded aluminium structures based on experimental results with various aluminium sheets in vehicle structures till load cycles $N = 10^7$ but not beyond this classical ultimate number of cycles. Bruder et al. [10] applied three different concepts, i.e. the nominal, the structural hot spot and the notch stress approach, with regard to applicability and quality of assessment. As an essential result of both studies it can be stated, that the fatigue assessment with existing guidelines based on the current state of the art may lead to a somewhat conservative design of weldments, hence partially contradicting the latest requirements in lightweight design, thus resulting in higher production costs and higher structural weight.

Recent investigations by Susmel et al. [11] on the fatigue behaviour of welded structures applied the notch stress approach in combination with the modified Wöhler curve method (MWCM). The MWCM allows a transformation of the conventional $S$–$N$ curve into a curve related to the maximum shear stress, which is assumed as fatigue life dominating maximum value under multiaxial loading conditions. The focus of this study was the assessment of fatigue behaviour of welded steel and aluminium structures especially under multiaxial loading conditions which could be applied successfully. For a fatigue life assessment beyond $N = 10^8$ the slope of the modified Wöhler diagram is hypothetically changed to $k = 22$ in accordance with the IIW-recommendations based on a suggestion made by Sonsino for the overall material group of aluminium alloys [12].

Studies on very high cycle fatigue of shot-peened welded joints made of Q345q steel at a stress ratio $R = -1$ at room temperature and ambient air were carried out by He et al. [13]. The results show that the fatigue strength of samples representing the weld seam are about 60% lower compared to the base material. It was also seen that the shot peening treatment significantly improved the fatigue strength of the welded samples. Inspection of the fracture surface revealed a crack initiation occurring in the base material at the surface below $10^7$ cycles until fracture and at subsurface voids in the VHCF range. For samples made of the weld seam the crack was initiated at inner defects. The shot peening process had no influence on the crack initiation site.

Recent findings on the VHCF behaviour of aluminium alloys demonstrate the diversity of fatigue life and related failure mechanisms in this load cycle regime [14–16,5]. Fig. 1a–d depict results for various aluminium alloys and leads to the assumption that a

![Fig. 1. S–N curves for (a) EN AW 6056 [14], (b) EN AW 7075-T6, EN AW 2024-T3 and EN AW 6061-T6 [15], (c) EN AW 6082-T6 [16] and (d) EN AW 2024-T351 [5].](image-url)
master S–N curve most likely will not do justice to the difference in fatigue life for the given alloys and heat treatment conditions in the VHCF regime. In particular the results for EN AW-6082 (Fig. 1c) show a strong microstructural influence on the VHCF behaviour, since the as-received condition shows a significantly superior cyclic strength compared to the over-aged condition. A comparison of fatigue life results under different environmental conditions (Fig. 1d) emphasises the critical aspect of testing procedure on the fatigue life data determined by means of high frequency fatigue testing. In general, cyclic strength for a number of load cycles \( N = 10^6 \) for the alloys given in Fig. 1a–d varies between \( \sim 80 \text{ MPa} \) and \( \sim 220 \text{ MPa} \). This discrepancy illustrates the necessity to further investigate the true nature of damage mechanisms dominating the VHCF behaviour of this material group and indicates that with increasing microstructural inhomogeneity, as can be found in welded structures, cyclic strength will all the more be dominated by microstructural effects, and fatigue life prediction will become even more challenging. Hence, a systematic study to characterise the influence of each (macroscopic and microscopic) notch effect at the different weld zones individually at very high cycles is the major motivation behind the investigation presented. An identification of the critical parameters dominating the VHCF behaviour of each zone serves as basis for recommendations for the welding design and process in order to optimise the cyclic strength.

2. Materials and methods

This investigation was carried out on metal inert gas (MIG) welded samples out of the precipitation-hardened aluminium alloy EN AW-6082 (AlMgSi1, AA 6082), combined with the filler material Al S 5183 (AlMg4.5Mn0.7, AA 5183) as well as artificially notched samples out of EN AW-6082. Both sample types were manufactured out of rolled blanks with a thickness of 6 mm for the welded and 20 mm for the notched specimens. The hour-shaped notched specimens were also extracted from rolled blanks in order to guarantee comparable conditions for both specimen constellations with regard to the rolling direction. Both specimen types were cut from the blanks so that loading and rolling direction are parallel. The chemical compositions and the quasi static mechanical properties of the materials used are given in Tables 1 and 2. Fig. 2a and b depicts the microstructure of EN AW-6082. According to Altenpohl [17] primary intermetallic precipitations with a size of several microns are formed during the casting process of the raw material. Fig. 2a depicts the microstructure of the EN AW-6082 with comparable precipitates of 2–25 \( \mu \text{m} \). Scanning electron microscopy (SEM) applying an X-ray energy dispersive analyser showed particles rich in silicon, manganese and iron (appearing bright) and particles rich in magnesium and silicon (appearing dark). According to [17,18] the bright precipitates containing Al (Mn, Si, Fe) are probably primary precipitates. These intermetallic particles are usually aligned linearly to the rolling direction and of a rather lath-like geometry. Comparable precipitates could be found in the studied material EN AW-6082 shown in Fig. 2b. The dark precipitates rich in Mg and Si are most likely large coherent second-phase particles with round shape and diameters from 2 to 6 \( \mu \text{m} \) also resulting from the initial alloy casting process.

<table>
<thead>
<tr>
<th>Table 1</th>
<th>Chemical composition [mass%] of the alloys studied.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Material</td>
<td>Mg</td>
</tr>
<tr>
<td>AlMgSi1 (EN AW-6082) 6 mm</td>
<td>1.030</td>
</tr>
<tr>
<td>AlMgSi1 (EN AW-6082) 20 mm</td>
<td>1.030</td>
</tr>
<tr>
<td>AlMg4.5Mn0.7 (Al S 5183)</td>
<td>4.800</td>
</tr>
</tbody>
</table>

Welded samples were prepared by means of a partially automated metal inert gas (MIG) welding process in the form of a double-Y butt joint. In order to define geometrical data for the definition of the artificially notched samples, the geometries of a total of 15 welded samples were measured. The average of each of the resulting parameters is given in Table 3. With these values the stress concentration factor \( k_t = 1.8 \) was calculated using the equation

\[
k_t = 1 + 0.27 \cdot (\tan \alpha)_{0.25} \frac{S}{R}^{0.5}
\]

based on the suggestion by Yung and Lawrence [19] with \( \alpha \) being the flank angle, \( S \) the plate thickness and \( R \) the radii of the weld toe. In addition to the macroscopic notch effect, typical process related weld seam imperfections such as hot cracks, pores and incomplete fusions which cannot be completely avoided even under laboratory processing conditions were also characterised in detail to correlate them with the fatigue behaviour and damage evolution. In this context the thickness of the aluminium oxide layer prior to welding and the specific combination of welding parameters (e.g. intensity of current, inert gas, etc.) are crucial features for the quality of the weld seam. Fig. 3a illustrates the area around the fusion zone. At the left side of the picture a weld seam (S Al 5183) with its typical cast structure consisting of dendrites can be seen followed by the fusion zone in the middle which represents the connection between the weld seam and the base material EN AW-6082 at the right side. According to Schulze [20] and Dilthey [21] imperfections like hot or liquation cracks mostly depend on the chemical composition of the welded material and the gradient of welding temperature. In the analysed welded joints no such defects were found but unwanted grain boundary segregation effects in the transition zone between weld seam and base material is observable (Fig. 3b). These unwanted microstructural changes are a result of remelted primary precipitates from the base material which accumulate at the grain boundaries and are highly disadvantageous in terms of monotonic and cyclic loading [18]. Frequently occurring gas pores resulting from hydrogen trapped in the solidifying weld seam can be seen in Fig. 3b. The major reason for the formation of these gas pores are usually an insufficient weld preparation or a wrong shield gas setting [20,21]. In the studied samples pores with sizes from 0.1 to 0.9 mm were found. The effect of temperature gradient across the weld seam during the welding process results in a change of monotonic strength across the weld seam due to the revoking of the original precipitation strengthening heat treatment, as is shown by hardness measurement across the weld seam seen in Fig. 4. These results represent average values within a scatterband of \( \pm 4 \text{ HV} \) for each single position in the different welding zones. The filler material Al S 5183 shows an approximately constant Vickers hardness of 85 HV marked at position I. In contrast, the precip-

<table>
<thead>
<tr>
<th>Table 2</th>
<th>Mechanical properties of the alloys studied.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Material</td>
<td>( R_{0.2} ) (MPa)</td>
</tr>
<tr>
<td>AlMgSi1 (EN AW-6082) 6 mm</td>
<td>297</td>
</tr>
<tr>
<td>AlSiMg1 (EN AW-6082) 20 mm</td>
<td>274</td>
</tr>
<tr>
<td>AlMg4.5Mn0.7 (Al S 5183) 0.15 mm</td>
<td>130</td>
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</table>
Fatigue tests were executed by means of an ultrasonic fatigue testing system under displacement control. Stress amplitude was defined by actual strain measurements by means of strain gauges prior to fatigue testing, based on Hook’s law and a Young’s modulus of $E = 71$ GPa. Tests were all performed at a stress ratio of $R = -1$. A detailed description of the basic principles of ultrasonic fatigue testing is given in [4,5]. Because of the special requirements of the resonating system, the geometries of the samples were simulated by means of the finite element software ABAQUS prior to fatigue testing. The sample geometries developed accordingly are given in Fig. 5. Sample geometries were restricted to welded samples with trimmed surfaces but the development of a specific calibration procedure for welded samples with remaining
Fig. 5. Ultrasonic fatigue sample geometry of the flat welded samples (a) and the different tested cross-sections consisting of base material (BM), weld seam (WS) and heat-affected zone (HAZ) (b) and the circumferential notched hourglass-shaped sample (c).

Table 4

<table>
<thead>
<tr>
<th>Heat treatment parameters</th>
<th>Heat treatment parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td>Material</td>
<td>Welding zone</td>
</tr>
<tr>
<td></td>
<td>III – (85 HV)</td>
</tr>
<tr>
<td></td>
<td>IV – (75 HV)</td>
</tr>
<tr>
<td></td>
<td>V – (110 HV)</td>
</tr>
<tr>
<td>Annealing</td>
<td>540 °C/60 min</td>
</tr>
<tr>
<td>Ageing</td>
<td>200 °C/60 min</td>
</tr>
<tr>
<td>Overageing</td>
<td>360 °C/7 min</td>
</tr>
</tbody>
</table>

reinforcement are currently under way. Due to the wave-like distribution of the load amplitude along the longitudinal axis of the UFTS samples, the critical cross-sections of the trimmed welded samples can either be positioned in the base material (BM), the weld seam (WS) or the heat-affected zone (HAZ) as is shown in Fig. 5a and b. Hence, the maximum stress amplitude is guaranteed to act in each individual welding zone if the position of the zone lies in the minimum of the cross section of the fatigue specimen. This way, the HAZ reflects the effect of temperature gradient in the welded joint on the originally peak-aged microstructure of the precipitation-hardening alloy EN AW-6082, whereas the weld seam additionally reflects the influence of process related weld defects. As the maximum stress amplitude extends to a length of approximately 3 mm, so does the critically stressed volume.

In addition to the welded samples, artificially notched hourglass-shaped specimens were used with a circumferential notch representing the same notch factor \( k_t \) as earlier defined for the weld seam with its toe radius, reinforcement and flank angle, see Fig. 5c. In order to define the static notch factor the analytical approach from the Forschungskuratorium Maschinenbau (FKM) e.V. guideline was used [7]

\[
k_t = 1 + \frac{1}{\sqrt{0.22 \cdot \frac{D}{t} + 2.74 \cdot \frac{t}{D} \cdot (1 + 2 \cdot \frac{t}{D})}}
\]  

(2)

with \( R \) being the notch radius, \( t \) the notch depth, \( D \) the external diameter and \( d \) inner diameter. The macroscopic notch effect from the weld reinforcement usually affects the stress concentration both in the heat-affected zone as well as in the weld seam. To take this into account the Vickers hardness of the artificially notched samples were adjusted by means of heat treatment in that way that the characteristic hardness values of the HAZ and the fusion zone were reached (Fig. 4). The corresponding heat treatment parameters are illustrated in Table 4. Notched base material samples (110 HV) were also included in order to evaluate the notch effect at different microstructures while surface condition, chemical composition and notch geometry are kept constant. In order to define the nominal and maximum notch stress in the cross section of the notched samples, strain is measured over the notch root by means of micro strain gauges. The measured strain values depend on the size, especially the length, of the strain gauge. The application of Hook’s law provides the average stress over the gauge length. The normalised stress distribution over the true path along the notch root was calculated by means of finite element method. Fig. 6 depicts this distribution indicating also the length of the micro strain gauge (1.4 mm) in the middle of the notch root. The modelling, meshing and the visualisation of the simulation were performed with ABAQUS, applying linear elastic material behaviour. By using the arithmetic average calculated with the function of the stress distribution over the gauge length from the simulation and setting it equal to the average stress determined on the basis of the measured strain and Hook’s law, the maximum notch stress (as indicated by the peak of the curve in Fig. 6) in the cross section can be estimated assuming linear-elastic material behaviour. A comparison of the nominal stress and the maximum stress of the notched samples confirms a static notch factor \( k_t \) (equal to 1.8). Any microplasticity effects are not reconsidered at that point but will be discussed later on. The surfaces of all specimens (trimmed weldments and artificially notched) were electrochemically polished. Metallographic and fracture surface investigations were conducted by means of optical microscopy (OP), scanning electron microscopy (SEM) and the image analysis software DATINF MEASURE. According to [22,23] the defect area was measured in the plane of observation which does not necessarily coincide with the plane of the largest section of the inclusion. However, the likely error is not expected to be large and is discussed in detail in [23].

3. Experimental results

3.1. S–N curve for welded samples

Fig. 7 presents the fatigue data for the three different weld zones, i.e. base material (BM), heat-affected zone (HAZ) out of
EN AW-6082 and the weld seam (WS) out of Al S 5183. For reason of comparison, the master S–N curve after Morgenstern [8] is also included. None of the weld zones shows a fatigue threshold beyond the classical fatigue limit ($N = 2 \times 10^8$), rather a continuous decrease of fatigue strength with almost the same values for BM as compared to the master S–N curve [8] could be observed. A slightly inferior cyclic strength for the HAZ as compared to the master S–N curve could be detected. The results for the BM also show a comparable cyclic strength with those reported by Höppel et al. [16] for the alloy EN AW-6082 in the peak-aged condition at the same stress ratio $R = -1$ (Fig. 1c). Fatigue results for the weld seam with run-outs at each stress level show a drastic decrease in cyclic strength and a pronounced scattering compared to the HAZ and BM, thus representing the weakest link of the weld joint in the VHCF range.

### 3.2. Fractographic investigations of BM and HAZ

Fig. 8a depicts a fracture surface of the base material ($\Delta \sigma/2 = 100$ MPa, $N_f = 1.52 \times 10^8$) showing void formation in the crack origin area. Fig. 8b documents that in the vicinity of void formation a rather large particle rich in Mg and Si (as proven by scanning electron microscope (SEM) with an X-ray energy dispersive analyser seen in Fig. 8c) and visible microcracks (at white arrows) can be found. The long and narrow shape and the large size of the particles rich in Mg and Si lead to the assumption that they are primary precipitates. Fracture surfaces of specimens representing the heat-affected zone which failed in the HCF regime ($N_f < 10^7$) clearly show crack initiation at the surface (Fig. 9a) with a typical feather-like surface structure. The crack initiation origin in HAZ samples that reached the VHCF regime ($N_f > 10^7$) cannot yet be allocated unambiguously (Fig. 9b).

### 3.3. Fractographic investigations of the weld seam and classification of weld defects

Fig. 10a depicts a surface crack already formed in the weld seam during cyclic loading in the HCF range. The crack path at the surface runs along a series of pores at the left side in positions I and II and towards an incomplete fusion in position III. Looking at the corresponding fracture surface of the same sample after final failure (Fig. 10b), it can clearly be seen that the crack initiation starts at position III representing the beginning of an incomplete fusion (marked with dashed line). Further propagation of the crack is supported by the pores in positions I and II and additional pores in the interior of the weld seam. The sample ($\Delta \sigma/2 = 60$ MPa, $N_f = 3.46 \times 10^7$) depicted in Fig. 10c shows similar weld defects as the previous sample. Here, position I can be identified as incomplete fusion (marked with dashed line) but without contact to the surface as seen before. This defect type with its rather narrow and elongated shape acts as crack initiation site.

Table 5 compares the number of loading cycles with the type and area of defects in the fracture plane for the VHCF range ($N_f > 10^7$). As a general trend a lower content of incomplete fusions in the cross-section of the specimens with increasing fatigue life can be observed. All run-out samples ($N = 2.1 \times 10^9$) show an accumulated defect area below 0.01 mm$^2$ supporting the previous statement. In sample 8 (compare Fig. 9d) a very huge pore ($\Omega = 1.2$ mm) is located in the centre area of the critical cross-section together with a very low amount of incomplete fusions. This sample reached a relatively high number of load cycles till failure ($N_f = 1.6 \times 10^8$) at a stress level $\Delta \sigma/2 = 70$ MPa. In contrast, samples tested at the same stress level but with higher content of incomplete fusions and smaller pores (max. $\Omega = 0.661$ mm), which were predominantly arranged in the form of centred clusters, al-

![Fig. 7](image-url) S–N data for welded samples with trimmed surfaces representing the different welding zones weld seam (WS), heat-affected zone (HAZ) and base material (BM) tested at a stress ratio $R = -1$ in comparison with the master S–N curve from [8] with a probability of survival $P_s = 97.7%$ (note: arrows denote run-outs).

![Fig. 8](image-url) (a) Void formation in the crack origin area of the base material in the VHCF region ($\Delta \sigma/2 = 100$ MPa, $N_f = 1.52 \times 10^8$), (b) detail of a micro void and adjacent intermetallic particle (black arrow) and microcrack formation (white arrows) and (c) chemical composition of the particle.
ready failed in a load cycle region between \(N_f = 7.9 \times 10^5\) and \(4.3 \times 10^6\).

3.4. Heat-treated and notched samples

One major disadvantage of the results presented hitherto is the disregard of the macroscopic notch effect due to the geometrical discontinuity of the weld reinforcement. As a consequence, artificially notched samples were tested in three different heat treatment conditions representing the fusion zone (85 HV), heat-affected zone (75 HV) and base material (110 HV) of the EN AW-6082. The nominal stress amplitudes and the load cycles till failure of these samples are shown in Fig. 11. The continuous and the dashed line represent the trend calculated by potential regression for smooth samples out of base material and notched samples with 110 HV (the nominal stress amplitude relates to the maximum stress amplitude as defined earlier on, reduced by the static notch factor \(k_t = 1.8\)). With increasing precipitation hardening strength the cyclic strength of the notched samples also increases. None of the different hardness conditions show a fatigue threshold beyond \(N_f = 10^6\), but a continuous decrease of cyclic strength. The notched peak-aged condition (with a hardness of 110 HV) shows a lower fatigue strength in the HCF and VHCF regime as compared to the results obtained for the smooth base metal samples. Results from Höppel et al. [16] on smooth samples out of EN AW-6082 in the three different heat treatment conditions show similar cyclic strength for the over-aged condition compared to the notched samples with 110 HV. In contrast, the as-received and the peak-

### Table 5

<table>
<thead>
<tr>
<th>Sample</th>
<th>(\Delta \sigma/2) (N/mm²)</th>
<th>(N_f)</th>
<th>Area pores (mm²)</th>
<th>Max. Ø (mm)</th>
<th>Area IF (mm²)</th>
<th>Area IF (%)</th>
<th>Area defects (%)a</th>
<th>Type and position of defect</th>
</tr>
</thead>
<tbody>
<tr>
<td>1  70</td>
<td>(7.9 \times 10^5)</td>
<td>0.373</td>
<td>0.186</td>
<td>1.214</td>
<td>5.06</td>
<td>6.61</td>
<td>Two huge IFs in centre</td>
<td></td>
</tr>
<tr>
<td>2  70</td>
<td>(1.2 \times 10^6)</td>
<td>1.287</td>
<td>0.251</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;5.36</td>
<td>IFs in centre surrounded by cluster of pores</td>
<td></td>
</tr>
<tr>
<td>3  65</td>
<td>(1.5 \times 10^6)</td>
<td>0.919</td>
<td>0.286</td>
<td>0.613</td>
<td>2.55</td>
<td>6.38</td>
<td>IF at surface</td>
<td></td>
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<tr>
<td>4  60</td>
<td>(1.8 \times 10^6)</td>
<td>2.001</td>
<td>0.379</td>
<td>1.294</td>
<td>5.39</td>
<td>12.99</td>
<td>IF at surface</td>
<td></td>
</tr>
<tr>
<td>5  70</td>
<td>(1.9 \times 10^6)</td>
<td>1.948</td>
<td>0.258</td>
<td>0.472</td>
<td>1.96</td>
<td>10.08</td>
<td>Pore at surface</td>
<td></td>
</tr>
<tr>
<td>6  80</td>
<td>(4.3 \times 10^6)</td>
<td>0.343</td>
<td>0.661</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;1.43</td>
<td>Pore at surface</td>
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<tr>
<td>7  60</td>
<td>(3.3 \times 10^6)</td>
<td>0.976</td>
<td>0.286</td>
<td>0.510</td>
<td>2.13</td>
<td>6.19</td>
<td>IFs in centre surrounded by cluster of pores</td>
<td></td>
</tr>
<tr>
<td>8  70</td>
<td>(1.6 \times 10^6)</td>
<td>5.801</td>
<td>0.925</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;15.80</td>
<td>Huge pore in centre</td>
<td></td>
</tr>
<tr>
<td>9  60</td>
<td>(4.2 \times 10^6)</td>
<td>0.014</td>
<td>0.068</td>
<td>0.067</td>
<td>0.28</td>
<td>0.34</td>
<td>Run-out, sporadic position of pores and IFs</td>
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</tr>
<tr>
<td>10 80</td>
<td>(2.1 \times 10^6)</td>
<td>0.283</td>
<td>0.014</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;1.17</td>
<td>Run-out, sporadic position of pores and IFs</td>
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<tr>
<td>11 70</td>
<td>(2.1 \times 10^6)</td>
<td>0.196</td>
<td>0.015</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;0.82</td>
<td>Run-out, sporadic position of pores and IFs</td>
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<tr>
<td>12 65</td>
<td>(2.1 \times 10^6)</td>
<td>0.251</td>
<td>0.012</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;1.05</td>
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<td>13 60</td>
<td>(2.1 \times 10^6)</td>
<td>0.120</td>
<td>0.053</td>
<td>&lt;0.01</td>
<td>&lt;0.04</td>
<td>&lt;0.54</td>
<td>Run-out, sporadic position of pores and IFs</td>
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</tr>
</tbody>
</table>

* Percentages are based on the nominal cross-section area of welded samples.
4. Discussion

4.1. S–N curve of welded samples

Experimental results of the present study prove a continuous decrease of fatigue strength in the VHCF range. While a prediction of fatigue life according to the master S–N curve and its extrapolation beyond the classical fatigue limit suggested by Morgenstern [8] leads to a rough evaluation for the BM of the welded structures investigated, VHCF strength of the HAZ and by far more drastically for notched samples in the three different Vicker’s hardness values are depicted in Table 6 and show a tendency of smaller fatigue notch factors at longer lives which was also found by Akiniwa et al. [25] in case of a bearing steel SUJ2 at very high cycles. With increasing hardness of the notched samples the fatigue notch factor decreases.

Table 6
Fatigue notch factors $k_f$ of notched samples with a Vicker’s hardness of 110, 85 and 75 HV at various cycles to failure.

<table>
<thead>
<tr>
<th>Load cycles</th>
<th>10^7</th>
<th>10^8</th>
<th>10^9</th>
</tr>
</thead>
<tbody>
<tr>
<td>$h_{(110HV)}$</td>
<td>1.47</td>
<td>1.45</td>
<td>1.42</td>
</tr>
<tr>
<td>$h_{(85HV)}$</td>
<td>1.65</td>
<td>1.62</td>
<td>1.47</td>
</tr>
<tr>
<td>$h_{(75HV)}$</td>
<td>1.87</td>
<td>1.76</td>
<td>1.62</td>
</tr>
</tbody>
</table>

Before analysing the pronounced decrease of cyclic strength for the WS by means of the fractographic results, the discrepancy in fatigue life between BM and HAZ must be discussed. As the weld reinforcement was trimmed due to the current requirements of the UFTS, the macroscopic notch effect can be ruled out, leaving the difference in microstructure as remaining influence factor. The base material represents the peak-aged condition (T6) of EN AW-6082 with a Vickers hardness of 110 HV. In this state of maximum hardness, inhomogeneities such as intermetallic inclusions or primary precipitations can lead to a localised stress concentration according to [15,16] inducing irreversible slip activities even beyond the classical fatigue limit. According to Wang et al. [15], who also studied an Al-alloy in peak-aged condition, during cyclic loading, this localised microstructural notch effect leads to the formation of so-called fatigue voids around these intermetallic particles which act as crack nucleation sites. Crack initiation begins with the formation and coalescence of these voids. According to the study of van Stone and Cox [26] voids in Al-alloys can nucleate at large second phases. However, this observation is solely related to monotonically loading condition. Sunder et al. [27] also observed void formation in AA-2024-T3 during cyclic loading. According to their investigations these voids were formed by fatigue induced interface decohesion between the secondary incoherent particles and the matrix. The voids serve as initiators for internal penny-shaped fatigue cracks, which grow to a noticeable size before they reach the surface or merge with another dominant crack. The voids and microcracks presented in this study (Fig. 8a–b) are comparable to those identified by Wang et al. [15] and Sunder et al. [27]. The analyses of the inclusions (Fig. 8c) lead to the assumption that in the case of the base material large primary precipitations are acting as inner crack initiation sites at very high cycles.

4.2. Fatigue behaviour of the base material

Atkinson et al. [25] carried out a study of the notch effect on the fatigue strength of a bearing steel SUJ2 in very high cycles. With increasing hardness of the notched samples the fatigue notch factor decreases. The results for the different precipitation conditions, as discussed later, clearly show that the microstructural influence cannot be neglected in the VHCF regime.

4.3. Fatigue behaviour of the heat-affected zone

The HAZ represents the over-aged condition of EN AW-6082 with an average Vickers hardness of 75 HV, implying a possible effect of the difference in precipitation morphology on the VHCF regime.
behaviour, as was shown for a precipitation-hardening nickel-based superalloy in the VHCF regime [28]. Höppel et al. [16] observed a pronounced microstructural influence on the VHCF behaviour for EN AW-6082 with an inferior cyclic strength of the over-aged condition compared to the peak-aged condition (Fig. 1c). Crack initiation was detected at the surface in the over-aged condition and peak-aged condition in the VHCF range. Pronounced slip markings at the surface were found. Höppel et al. [16] assumed that a rather soft matrix of both heat treatment conditions promotes irreversible cyclic slip more effectively compared to an as-received condition with a harder matrix and hence the plastic deformation will be localised at the surface resulting in the formation of cyclic slip markings. These findings are a promising approach to explain the difference in cyclic strength of the base material (110 HV) compared to the softer HAZ (75 HV) presented in this study.

4.4. Fatigue behaviour of the weld seam

The fractographic analyses of the WS indicate that crack initiation mostly starts at incomplete fusions and fracture paths favoured porous areas for subsequent propagation. Hence, the rather sharp and narrow contour of the incomplete fusion is much more effectively acting as a local stress raiser as the smooth round shape of a pore (Fig. 10c) and the stress concentration seems to be the major reason for the earlier failure of these samples. In a first attempt to evaluate the huge scatter of these fatigue results, the square root area model by Murakami was applied. The \( \sqrt{\text{area}} \) for incomplete fusion was determined on the basis of the recommendations of Murakami [23] for narrow elliptically shaped non-metallic inclusions with \( (l/c) > 10 \)

\[
\sqrt{\text{area}} = \sqrt{10 \cdot c}
\]

(4)

with \( \sqrt{\text{area}} \) is the square root of the area, \( c \) is the height and \( l \) is the length of the incomplete fusion. In case of crack initiation at a pore, the classical Murakami model with the correction method for non-ferrous metals according to Noguchi et al. [29] was used

\[
\sigma_w = \frac{\alpha \cdot (HV + 120 \cdot \frac{E_{Al}}{E_{St}})}{\sqrt{\text{area}}}^{1.6}
\]

(5)

Here \( HV \) is the Vickers hardness, \( E_{Al} \) is the Young’s modulus for aluminium (here: \( E_{Al} = 71 \) GPa), \( E_{St} \) the Young’s modulus for steel (here: \( E_{St} = 210 \) GPa) and \( \alpha \) was set to 1.43 for surface defects and 1.56 for inner defects. A good correlation between normalised stress ratio \( (\Delta\sigma/2)/\sigma_w \) and the number of load cycles until failure (Fig. 12) is observed only in the VHCF region both for crack initiation at pores and at incomplete fusions. By contrast, the value for the normalised stress ratio of sample 8 with a large inner pore (compare Table 5 and Fig. 9c) is about 12 and, hence, deviates drastically, probably indicating the limits of validity of the model for welding defects. For reasons of a better distinction of the other results, the estimated result for sample 8 is not included in the diagram shown in Fig. 12. Results of the run-out samples are very close to zero which stands in good correlation with the measured very small area of defects (see Table 5). However, only the geometry of the crack initiating defect is considered in this approach. Furthermore the Murakami model was developed to describe the influence of small defects. Therefore the application of the Murakami model can only be seen as preliminary approach to classify the influence of defect size for the VHCF behaviour of the WS. A more differentiated analysis of the correlation between fatigue life and welding defects will also have to take the position of failure relevant defects as well as the overall defect distribution in the critically loaded volume fraction of the fatigued samples into account.

4.5. Heat-treated and notched samples

The increase of cyclic strength of notched samples with increasing precipitation hardening, i.e. the difference in cyclic strength of the three conditions representing the different weld zones (75, 85 and 110 HV) can be observed in Fig. 11. The influence of the microstructure on the cyclic strength can clearly be proven while the notch geometry, surface condition and chemical composition of the specimens used were identical. The so called microplasticity effect, which has not yet been characterised for the results presented, has to be taken into consideration whenever discussing the cyclic strength of notched samples. The microplasticity effect is a material and monotonic strength dependant parameter (see [30,31]), representing the effect of microstructure on the true (rather irregular) stress distribution in the notch vicinity resulting in a higher deformation resistance than predicted on the basis of linear-elastic material behaviour (for a detailed overview of the definition of the microplasticity effect see [8]). The difference in fatigue strength in the VHCF regime of the notched specimens might be explained by the precipitation hardening condition and its influence on this effect. The decreasing fatigue notch factor \( k_f \) (see Table 6) for different Vicker’s hardness values underlines this assumption for the notched samples.

The fatigue data and microstructural analyses of the study presented are the result of a first systematic investigation of the VHCF properties of a welded precipitation hardened aluminium alloy by means of an ultrasonic fatigue testing system. The difference in VHCF behaviour of the weld zones could be correlated to its specific microstructural features and welding defects, respectively. However, for a reliable lifetime prediction model, as is needed for safety relevant welded structures, for the VHCF regime, further studies are needed. The results presented demonstrate the complexity of fatigue damage behaviour in the VHCF regime. The cyclic strength of inhomogeneous structures as investigated in the study presented cannot be reduced to a single influence factor such as the macroscopic notch effect. On the one hand a clear microstructural influence could be demonstrated for notched samples. On the other hand, the influence of local notch effects due to incomplete fusion, pores or other welding defects, as was shown for the welded samples, will become all the more important the lower the overall stress amplitudes. In order to establish a fatigue life prediction approach on the basis of the true nature of failure mechanisms, the results emphasise the diversity of such mechanisms, “exploding” with increasing microstructural inhomogeneity. Hence, not only the integral fatigue behaviour of the overall welded structure but also the damage mechanisms of the underlying microstructure have to be further investigated.
In this respect, further investigations concerning the VHCF behaviour of the base material, such as the influence of the secondary incoherent precipitates or the dislocation/particle interaction should be pursued. This might serve as basis to explain the difference in microplasticity effect for the different heat treatment conditions. It might also serve as basis to evaluate the likely influence of precipitate segregations sometimes observed near the fusion zone of welded aluminium structures.

The question whether fatigue life is dominated by crack initiation or fatigue crack growth with regard to the defect type cannot be answered yet. In this respect, future investigations applying indirect crack growth measurements techniques such as the potential drop method or the nonlinear ultrasonics technique will help to elucidate the fraction of crack initiation versus crack growth for the defect afflicted weld seam.

5. Conclusions

Ultrasonic fatigue tests were carried out with samples extracted from double-Y butt welded sheets and artificially notched samples in the peak-aged and two different over-aged heat treatment conditions representing the different zones of the welded sheets. By trimming the surfaces of the welded samples and varying the position of the highest stressed region of the ultrasonic fatigue samples relative to the weld zones, the microstructural and the macroscopic notch effects could be evaluated separately. The main results can briefly be summarised as follows:

- Ultrasonic fatigue tests of welded samples with trimmed surfaces show that the weld seam is the weakest region compared to the heat-affected zone and the base material in the HCF and VHCF region. Fatigue behaviour is dominated by crack initiation and subsequent growth caused by weld inhomogeneities (incomplete fusions and gas pores). The heterogeneous distribution of these defects in combination with the small testing volume results in a huge scattering of fatigue data. Fracture surface analysis indicated a tendency towards earlier crack initiation at incomplete fusions than at gas pores. The square root area model by Murakami to evaluate the effect of defect size and geometry shows a reasonable predictable capability in the VHCF regime.

- An extrapolation of the master S–N curve suggested by Morgenstern [8] serves as suitable basis for the prediction of VHCF life of the base material. However, fatigue strength of the heat-affected zone, the weld seam and the notched samples with 85 and 75 HV is explicitly lower than described by the master S–N curve. Results of the present study could experimentally verify the continuous decrease of fatigue strength in the VHCF range which was only assumed before.

- In the VHCF regime fatigue voids were found in the vicinity of the crack origin for samples representing the base material. On the basis of findings of other authors who also investigated precipitation hardened aluminium alloys, it can be stated that the formation of voids is probably a consequence of interfacial decohesion of incoherent precipitations rich in Mg and Si resulting in a subsequent coalescence of microcracks and eventually leading to failure relevant crack growth. However, the inferior fatigue strength in the VHCF regime of the over-aged condition with 75 and 85 HV compared to the base metal also gives rise to the possible influence of the interaction between dislocation motion and the size, shape and distribution of coherent/semi-coherent precipitates adjusted by heat treatment prior to fatigue testing.

- Fatigue results for artificially notched samples (with a stress concentration factor similar to that of the weld reinforcement) in different heat treatment conditions show a clear microstructural effect manifesting itself in a decrease of the fatigue strength with decreasing hardness. A comparison of the results with those related to smooth samples of correspondingly heat-treated conditions was possible, as the nominal stress amplitudes for the notched samples are based on strain measurements in the notch toe, a calculation by means of finite element method and the static notch factor \( k_t \). The discrepancy between notched and smooth sample results might partially be a result of microplasticity effects and its dependence on the static strength of the different heat treatment conditions. In addition, a decrease of the fatigue notch factor \( k_t \) with increasing load cycles till failure could be detected.

The results presented demonstrate the complexity of the fatigue damage behaviour in the VHCF regime. At the given state of investigation no conclusive recommendation for a reliable fatigue life assessment of welded aluminium joints can be expressed for the VHCF range. However, the difference in VHCF behaviour of the weld zones could be correlated to its specific microstructural features and welding defects, respectively. Fatigue life prediction models that are solely based on the macroscopic notch effect do not justice to the complexity of damage mechanisms observable in the VHCF range. Hence, the results presented serve as sound basis for a better understanding of the failure of welded structures in the VHCF regime. Future investigations on welded samples with reinforcement and varying weld quality will have to prove, whether the pronounced influence of welding defects will remain the life dominating feature and whether the Murakami model is a suitable basis to define the minimum necessary weld quality for a beneficial VHCF behaviour.

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References