Fatigue behaviour and fractography of extruded AZ80 magnesium alloys in very high cycle regime

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Abstract

In order to investigate the fatigue properties of extruded magnesium alloy in very high-cycle regime, rotary bending fatigue test was performed in ambient atmosphere at room temperature using the hourglass shaped specimens of AZ80 alloys extruded (F-specimen) and treated by an artificial aging after extrusion (T5-specimen). From the experimental results, both specimens show a clear step-wise S-N curve on which two knees exists. Specific stress amplitude formed the knee corresponded to the 0.2% offset proof stress of 160MPa in compression. From the detailed observation of fracture surface, small facet-like smooth area appeared at crack initiation site on the specimen failed at low-stress amplitude level and in a high-cycle regime. On the other hand, lamella structure appeared at crack origin in case of the test at high-stress amplitude level. It was concluded that the step-wise S-N curve was induced by the crack initiation mechanism changing from the twin deformation in high-stress amplitude level to slip deformation in low-stress amplitudes, because of a mechanical anisotropy caused by the extrusion and the particular deformation characteristics of the hexagonal crystal structure in a magnesium alloy.

Keywords: Very High Cycle Fatigue; Extruded Magnesium Alloy; Step-wise S-N Curve; Fractography; Crack Initiation; Twin; Artificial Aging

1. Introduction

From the viewpoint of high efficiency, saving energy, minimizing the weight of structural elements and reduction of the environmental loads to the globe, magnesium alloys, having some excellent properties such as low density, high specific strength and stiffness, good machine-ability and so on, have received more concerns as very attractive materials for structural components, especially traffic-body, power transmission and engine parts. With development of extruding and rolling technology in present-day, wrought magnesium alloys are expected to use of mechanical components under high oscillating load, instead of using cast magnesium alloys which have disadvantages in defects such as casting porosity, cavity and microscopic shrink hole. For application of the wrought alloys to load-bearing components, it is important to evaluate fatigue properties. Some experimental studies have been carried out about high cycle fatigue properties and fatigue crack propagation behavior. Through these studies, it was not clear to exist the fatigue limit on S-N curve of extruded magnesium alloy up to $10^7$ cycles. Recently, Yang et al. has been reported

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from an ultrasonic fatigue testing that the S-N curve of extruded AZ31 appeared to have a trend of continuous decreasing without a horizontal asymptote up to $10^7$ cycles [1].

It is already recognized for magnesium alloys to result a strong crystallographic anisotropy and texture, because of the limited number of active slip systems and then operations of twinning systems in hexagonal close packed lattice structure [2-8]. It is well known that the main deformation systems for hcp-structured magnesium alloys at room temperature are both \{0001\} \langle11-20\rangle basal slip and \{10-12\} twinning. Also, extruded magnesium alloy has a strong fiber texture with \{0001\} basal plane aligned parallel to the extrusion direction. To evaluate fatigue properties of the wrought magnesium alloys, problem encountered is the strong anisotropy of mechanical properties, which is caused by both the crystallographic nature and the mechanical processing. It is important that both cyclic slip deformation and twinning are taken into account to discuss the mechanism of crack initiation and propagation in high cycle fatigue. But, knowledge of the influence of twinning on fatigue behavior is still lacking.

The aim of this study is to discuss very high cycle fatigue properties of extruded magnesium alloys. An experimental investigation was performed under rotary bending fatigue test using hourglass-shaped specimen of commercial extruded AZ80. Because that the specimen showed specific step-wise S-N curve, mechanism of surface fatigue crack initiation was discussed through the detailed observation of fracture surface with scanning electron microscope (SEM). In addition, the effect of an artificial aging after extrusion on very high cycle fatigue behaviors was examined.

2. Experimental Procedures

2.1. Testing materials and specimen

The material used in this study was commercial Mg-Al-Zn magnesium alloy, AZ80. The chemical composition of these materials (in mass percentage) is 8.24Al, 0.67Zn, 0.20Mn, 0.005Fe, 0.012Si, 0.0008Cu, 0.0007Ni and balanced Mg. The bar with 16mm in diameter was extruded from a billet of 160mm diameter (an extrusion ratio of 99.4:1) under the extrusion ram speed of 1.5m/min at temperature of 623K. Hour-glass shaped specimens with a grip diameter of 10 mm and minimum diameter of 5 mm (Fig. 1) was machined from the extruded bar with the loading axis parallel to their extrusion directions. The elastic stress concentration factor, $K_t$, of these specimens was 1.065. On the other hand, to investigate the effect of an aging treatment on fatigue behavior, specimens were prepared from the extruded bar which was heated at 473K for 32h and then air-cooled (T5 treatment). From now on, the specimen extruded and no heat-treated is referred to as AZ80-F (F-specimen), and the specimen artificially aged as AZ80-T5 (T5-specimen). To avoid the influence of surface roughness on the fatigue behavior results, the round-notched surface was polished along the longitudinal direction with a series of emery papers up to mesh of #2000 and subsequently finished by buff polishing before fatigue testing.

Figure 2 shows the optical micrographs illustrating the microstructure of the tested materials. The grain structure was clearly observed on the F-specimen (a), and the morphology observed on the vertical plane to the extruded direction/specimen axis was same as that on horizontal one. The average grain size is 16.0μm. On the other hand, the grain structure could not be clearly observed on the T5-specimen, because of the intermetallic phase (β-phase, Mg$_{17}$Al$_{12}$) precipitated by the aging. Observation of the precipitated microstructure of T5-specimen is shown in Fig. 2 (b) and (c). The intermetallic phase is formed columnar on the α-matrix along the extruded direction as illustrated with the optical micrographs, and shows lamellar pearlite-like structure from the detailed observation with SEM (d). Vickers hardness of the β-phase and the α-phase is 85.5HV and 73.3HV, respectively.

![Fig. 1. Shape and dimensions of tested specimen (mm).](image-url)
The mechanical properties were summarized in Table 1. Proof stress (0.2% offset), ultimate tensile strength obtained from monotonic tensile test and average Vickers hardness are increased, and the elongation is slightly decreased by the T5-aging treatment as compared with those of F-specimen. Figure 3(a) and (b) show stress vs. strain curves obtained from monotonic tensile test and incremental-step loading and unloading test (ISLUT) \cite{9, 10} for F- and T5-specimen. Stepwise strain of about 0.1% increment was applied in tension or in compression after unloading. Proof stresses in tensile and compressive side measured from the ISLUT are summarized in Table 1. The compressive proof stress is obviously lower than the tensile one. The difference between the proof stresses in F-specimen is larger than T5-specimen. Asymmetry of proof stress between in tension and compression will be caused by a strong fiber texture due to the extrusion with the c-axis distributed perpendicular to the extrusion direction. The \{0002\} pole figure shown in Fig. 3(c) confirms that \{0001\} basal plane distributes parallel to the extrusion direction and both the \{0001\}<11-20> basal slip and \{10-12\} twinning are difficult to act under tensile loading, while \{10-12\} twinning under compressive loading is extensively favorable to operate in such orientation.

2.2. Fatigue testing method

The fatigue tests were performed in an open environment at room temperature using a four-axis cantilever-type rotary bending fatigue machine, which can simultaneously perform the test four different specimens at a frequency
of 80Hz. Fracture surface was observed by scanning electron microscope (SEM) and quantitative analysis was conducted with computer-aided three-dimensional image processing equipment.

3. Experimental Results and Discussion

3.1. S-N curve characteristics

Figure 4(a) shows the S-N curve obtained from respective F- and T5-specimens with a testing cycle range up to $10^9$ cycles. From the examination of fracture surfaces by SEM as mentioned below section, it was observed that the fatigue failure originated from the surface of specimens for all range of stress amplitude and fatigue lifetime. The S-N curve of two specimens appears to have a trend of continuous decreasing at high stress amplitude range, 220–160 MPa, and short fatigue lifetime below $10^5$ cycles. The difference in fatigue strength between two types of specimen cannot be found. At the stress amplitude level around 160-150MPa, the S-N curve shows horizontal tendency and thereafter continuously decreasing trend; above range of $10^6$ cycles for T5-specimen and $2\times10^7$ cycles for F-specimen. Fatigue strength of the T5-specimen is lower than that of the F-specimen at low-stress amplitude level and high number of cycles, even though higher static mechanical properties with the aging treatment. Therefore, both specimens show a clear stepwise S-N curve on which two knees exist. Fatigue strength at $10^9$cycles is 135MPa for the F-specimen and 110MPa for the T5-specimen, and fatigue limit is not observed for the materials within a tested cycle range. Fatigue strength ratio, $\sigma_a/\sigma_B$, at $10^9$cycles is 0.44 for the F-specimen and 0.32 for the T5-specimen (Fig.4(b)), and these values are appropriate to previously reported value of 0.25–0.5 for wrought alloys, but the tendency contradicts the fact that higher fatigue strength ratios are for higher strength alloys.

3.2. Characteristics of fracture surface

3.2.1. Low stress amplitude level

Figure 5 shows a typical fracture surface of the T5-specimen tested under low-stress amplitude level ($\sigma_a=150$MPa, $N_f=1.67\times10^6$) observed with SEM. From low magnification image in Fig. 5(a), the fracture surface around crack initiation site can be divided into three regions, that is, dark area at crack origin, bright and white area (referred to as region IIB) outside the dark area and gray area surrounded the region IIB (referred to as region IIC). Specific morphology of lamellar structure was observed in region IIB with higher magnification SEM image (b). The dark area is also divided into two regions as shown in Fig. 5(c). Flat and facet-like area exists at crack origin as a result of observation with higher magnification (d) and will be called as region I. Area outside the region I (referred to as region IIA) shows a morphology having river-pattern like stripes which spread to radial directions from crack origin. In this region, the lamellar structure observed in region IIB could not be found.

In order to discuss the morphology of the fracture surface in detail, surface roughness measurements of the
The fracture surface along a cutting plane line, AB, shown in Fig. 5(c) were taken by three-dimensional SEM analysis. The surface roughness curves are shown in Fig. 6. Region I has a tilt to about 45 degree along the direction of applied normal stress / longitudinal specimen axis, and the slope changes toward gradually vertical direction of applied normal stress in the region IIA. The surface roughness of the region IIB is greater than that of the region IIC.

3.2.2. High stress amplitude level
Observation of typical fracture surface obtained from the F-specimen (σ_a=180MPa, N_f=8.98x10^4) ruptured at the level of high-stress amplitude and short fatigue lifetime was shown in Fig. 7. Region I, flat and facet-like area, and region IIA cannot be found around crack origin contradictory to the fracture surface ruptured at low-stress amplitude level. Morphology of the region IIB exists around crack initiation site.

3.2.3. Dependency of stress amplitude level on the morphology of fracture surface
Table 2 shows values of roughness measurement, R_a and R_max, along the cutting plane line of fracture surface obtained from both specimens tested at high and low-stress amplitude. An average distance of roughness on surface roughness curve, S, is also shown in this table. Roughness on region IIB of both specimens is larger than those on
Fig. 7. An example of SEM observation on fracture surface of AZ80-F ruptured at short fatigue lifetime: (a) Macroscopic view around the crack initiation site and (b) Higher magnification of crack origin.

Table 2. Roughness measurements along the cutting plane line of the fracture surface as shown in Fig. 5(c).

<table>
<thead>
<tr>
<th>Material</th>
<th>Stress amplitude</th>
<th>Roughness</th>
<th>Region</th>
<th>Region</th>
<th>Region</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>IIA</td>
<td>IIB</td>
<td>IIC</td>
</tr>
<tr>
<td>AZ80-F</td>
<td>$\sigma_a=180\text{MPa}$</td>
<td>Ra</td>
<td>2.927</td>
<td>1.520</td>
<td></td>
</tr>
<tr>
<td></td>
<td>$N_f=8.98\times10^4$</td>
<td>Rmax</td>
<td>12.827</td>
<td>6.745</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>S</td>
<td>8.645</td>
<td>7.356</td>
<td></td>
</tr>
<tr>
<td>AZ80-T5</td>
<td>$\sigma_a=140\text{MPa}$</td>
<td>Ra</td>
<td>2.943</td>
<td>1.586</td>
<td></td>
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<tr>
<td></td>
<td>$N_f=4.88\times10^4$</td>
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<td>12.196</td>
<td>6.780</td>
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<tr>
<td></td>
<td></td>
<td>S</td>
<td>6.273</td>
<td>6.847</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 8. Experimental results of stress intensity factor range at the boundary of region I, IIA and IIB on the fracture surface.

region IIA and IIC, and roughness on each region does not show the dependency on stress amplitude level. It is considered that the greater roughness on region IIB resulted from the lamellar structure observed on the fracture surface. From the roughness measurement along the vertical direction to the lamellar structure, $R_{\text{max}}$ on region IIB was $1.2 \sim 1.9\mu\text{m}$ and $S$ was $3.4 \sim 3.5\mu\text{m}$, and these values were same as region IIC. Average unit of lamellar structure on the fracture surface, which is formed into same direction, is $15.8 \sim 15.9\mu\text{m}$ for both specimens and same as the grain size. Also this value did not depend on the applied stress amplitude level.

Area of each region on fracture surface was measured and stress intensity factor range, $\Delta K$, was calculated using Equations (1) and (2).

$$\Delta K_{\text{I,IIA}} = 0.65\sigma_a\sqrt{\text{area}_{\text{I,IIA}}}$$

for region I and IIA
where, $F$ in Eq. (2) is a parameter depending on aspect ratio of the area in region IIB [11]. Figure 8 shows the experimental relationship between $ΔK$ in each regions and number of cycles to failure, $N_f$, for F- and T5-specimen. From this figure, $ΔK_{I}$, $ΔK_{IIA}$ and $ΔK_{IIB}$ changing from region I to IIA, IIA to IIB and IIB to IIC, respectively, show the constant value not depending the $N_f$.

3.3. Fatigue crack growth behavior

Morphologies of crack initiation site and crack tip on a specimen surface were observed with SEM. A typical example is shown in Fig. 9. Crack of the F specimen tested at $\sigma = 180$ MPa is initiated in the grain and propagated in the interior of the grain with changing the direction at the grain boundary as shown in Fig. 9 (a) and (b). Twin boundaries were found in the grain and crack is formed along the twin boundary. From Fig. 9 (c) and (d) for the T5 specimen tested at low-stress amplitude of $\sigma = 140$ MPa, crack is initiated in the $\alpha$-matrix and propagated in the precipitated intermetallic phase by a zigzag path.

Surface crack growth behavior during fatigue process was observed by the replica method taken from the specimen surface interrupted at intervals for measurement. Figure 10 shows the experimental measurement of surface crack length relating to fatigue life ratio, $N/N_f$. Fatigue crack initiates at an early stage of fatigue cycling, $N/N_f=0.1–0.2$, and grows continuously under cycling at high stress amplitude level (a). On the other hand, at low stress amplitude level, the crack initiated at an early stage of fatigue process propagates very slowly for the major part of a fatigue life and the crack grows quickly at the range in 10% of the remained fatigue life (b). Small crack formed in weak grain on specimen surface by cyclic slip deformation at low stress amplitude level may be difficult to grow toward neighboring grains, because of a lack of slip systems for plastic deformation at a crack tip. Much number of cycling is required after forming the region I, and the propagation of crack is very slow at this time and then a region IIA generates. After formation of an appropriate size for the region IIA to enable propagation as an ordinary crack, the crack propagates with forming a region IIB. It is suggested that crack at high stress amplitude

\[
ΔK_{IIB} = F\sigma_a \sqrt{\frac{\pi}{2} \text{area}_{IIB}}
\]

for region IIB

Fig. 9. Morphology of crack initiation site and crack tip of AZ80-F tested at $\sigma = 180$ MPa (a and b), and AZ80-T5 tested at $\sigma = 140$ MPa.
level initiates as a result of deformation twin formed during compressive loading. The twin boundary could act as preferential sites for crack nucleation. The sharp edge of existing twin lamellas causes a local stress concentration and a region IIB is observed at crack initiation site as shown in Fig. 7.

3.4. Mechanism of fatigue failure

Difference in the proof stress between tensile and compressive side will affect on fatigue damage, such as fatigue deformation and crack initiation and propagation. Although the maximum stress is below the proof stress in tension under the completely reversed tensile to compressive loading fatigue, it is possible that the minimum stress exceeds the proof stress in compression and twinning deformation forms under compressive loading. The activation of twinning mode will be led only by compressive loading but not tensile one to perpendicular the c-axis. The remaining interfaces between twin and matrix during fatigue cycling could act as preferential sites for crack initiation [1]. Figure 10 shows the experimental relationship between the minimum stress (stress amplitude) divided by the proof stress in compression of 150MPa for F-specimen, \( \sigma_{\text{min}}/\sigma_{0.2} \), and \( N_f \). Experimental data in this figure were plotted by the classification of two groups with the observation of fracture surface as has been mentioned in section 3.2. One is the specimen observed a flat and facet-like area, region I, at crack origin and indicated with a solid mark. The other is that observed a lamella structure, region IIB and not existed Region I at crack origin, and indicated with an open mark. From this figure, two groups are clearly separated each other at the level of \( \sigma_{\text{min}}/\sigma_{0.2} = 1 \). In low-cycle regime above this level, it is considered that irreversibility of cyclic deformation caused by twinning is main factor influencing the crack initiation. On the other hand, cyclic slip could act to initiate a crack in a high-cycle regime below the level at which twinning are difficult to operate under compressive loading. The
conventional horizontal line showed in the S-N curve may be a limit for fatigue crack initiation and propagation controlled by twinning deformation, that is, a certain level led to a transition of dominant deformation mode from twinning to slipping. Above discussion is explained for not only F-specimen but also T5-specimen having high compressive proof stress, because that a fatigue crack initiates in α-phase structure of T5-specimen.

4. Conclusions

(1) The specific stepwise S-N curve existing two knees appeared for both specimens extruded and treated by a T5-artificial aging after extrusion under the rotary bending fatigue test in the very high cycle regime.

(2) The knee, conventional horizontal line in S-N curve, appeared at stress amplitude of 150-160MPa which corresponds to 0.2% offset yield stress in compression of the material.

(3) Fracture surface was classified into four characteristic morphologies excluding a final fracture region, which depend on applied stress amplitude level. The transition of the fracture region is decided uniquely by stress intensity factor range.

(4) Fatigue crack initiation could be controlled by irreversibility of twinning at high-stress amplitude level, and changed to mechanism of cyclic slip deformation at low-stress amplitude and in very high cycle regime.

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References