Fatigue behaviour and fracture mechanism of a rolled AZ31 magnesium alloy

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Abstract

Axial fatigue tests have been performed using smooth specimens and CT specimens of a rolled AZ31 magnesium alloy in laboratory air at ambient temperature. Fatigue strength and fatigue crack propagation (FCP) characteristic were evaluated and fracture mechanism was discussed on the basis of crack initiation, small crack growth and fracture surface analysis. The relationship between FCP rate and stress intensity factor range for large cracks consisted of two sections with different slopes due to the transition of the operative micromechanisms of fracture. The FCP resistance was significantly lower than that of aluminium alloys and pure titanium. The fatigue strength at 10^7 cycles was 50 MPa that led to a considerably low fatigue ratio of 0.23. There existed two different modes of crack initiation depending on applied stress level. Above 70 MPa, cracks initiated at the specimen surface in transgranular or intergranular manner due to cyclic slip deformation, while below that stress subsurface crack initiation took place. The growth of small cracks initiated at the surface coincided with the FCP characteristic after allowing for crack closure for large cracks, but the operative fracture mechanisms were different between small and large cracks. At the subsurface crack initiation site, smooth facets were always present regardless of applied stress level. The features of the facets were very similar to those observed in titanium alloy that have the same crystal structure. The facet sizes were smaller than, or nearly the same as, the average grain size, suggesting a crystallographic nature of the facets. Furthermore, there was no correlation between the maximum stress intensity factor for facet and fatigue life.

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1. Introduction

Magnesium (Mg) alloys are very attractive materials in order to achieve high performance and energy saving of machines and structures, because of their advantages such as light weight, high strength-to-weight ratio and high specific stiffness. Therefore, they have recently been increasing interest as structural materials in many applications, in particular, they are considered to be replacing aluminium alloys with a concomitant saving in weight in automotive industries. For applications to load-bearing components, it is necessary to evaluate various fatigue properties.

Ogarrevic and Stephens have reviewed the fatigue data of Mg alloys published between 1923 and 1990 and indicated that a significant amount of fatigue strength data existed, but most of it was not recent and FCP data were mostly obtained in USSR [1]. When considered such situation of fatigue research and the recent development of Mg alloys, the accumulation of various fatigue data is now of particular importance. Mg alloys can be classified into two categories, casting and wrought alloys. In casting alloys, defects such as casting porosity and cavity are usually present and the fatigue properties are affected significantly by their shape and dimension [2,3]. On the contrary, wrought
alloys are basically defect-free and thus the evaluation of their fatigue properties is of great interest to understand the intrinsic fatigue mechanism of Mg alloys. However, there have been limited studies on wrought Mg alloys [4,5].

In the present study, axial fatigue tests have been performed using smooth specimens and CT specimens of a rolled AZ31 Mg alloy in laboratory air at ambient temperature. Fatigue strength and FCP characteristic were evaluated and fracture mechanism was discussed on the basis of crack initiation, small crack growth and fracture surface analysis.

2. Experimental details

2.1. Material

The material used is a commercial AZ31 Mg alloy rolled plate with a thickness of 6 mm. The chemical composition of the present alloy is not known, but according to the material standard, Al 3% and Zn 1%. The material was provided in as-rolled condition and no heat treatment was applied to fatigue specimen before experiment.

Fig. 1 shows the microstructure of the alloy, which consisted of equiaxed grains and macroscopic anisotropy of the microstructure is not recognized. The average grain size is approximately 60 µm. The mechanical properties in the rolling direction are 0.2% proof stress $\sigma_{0.2}$: 110 MPa, tensile strength $\sigma_B$: 224 MPa, elongation $\phi$: 30%.

2.2. Specimens

As shown in Fig. 2, fatigue specimens with 8 mm width, 12 mm gauge length and 4 mm thickness were machined from as-received plate so that their axis is parallel to the rolling direction. A blunt notch whose stress concentration factor is 1.03 was introduced on one side of the specimen to facilitate observation of fatigue crack initiation and small crack growth. Specimens were mechanically polished by emery paper and then buff-finished before fatigue test.

CT specimens with 50.8 mm width and 6 mm thickness were also prepared from as-received plate. The FCP direction was perpendicular to the rolling direction, i.e. L–T orientation.

2.3. Procedures

Fatigue strength data were obtained at a stress ratio, $R$, of ~1 using an electro servohydraulic fatigue testing machine operating at a frequency of 10 Hz in laboratory air at ambient temperature. Crack initiation and small crack growth were monitored with replication technique. Fatigue testing was periodically interrupted and replicas were taken and then crack length was measured using a laser microscope.

FCP tests using CT specimens were performed at $R = 0.05$ using the same fatigue testing machine and frequency as employed in fatigue strength test. A fatigue pre-crack of 2 mm long from the notch root was introduced and then increasing or decreasing stress intensity factor range, $\Delta K$, tests were started. In decreasing $\Delta K$ tests, a load-shedding technique was employed. Crack length was monitored with a traveling microscope with a resolution of 10 µm. Crack closure was measured by a compliance method using a strain gauge mounted on the back face of the specimen.

After the experiment, fracture surfaces were examined in detail by scanning electron microscope (SEM).
3. Results and discussion

3.1. FCP characteristics for large cracks

The relationships between crack propagation rate, \( da/dN \), and \( \Delta K \) and effective stress intensity factor range, \( \Delta K_{\text{eff}} \), for large cracks are represented in Fig. 3. It is worth noting that the \( da/dN-\Delta K \) relationship consists of two sections with different slopes below and above \( \Delta K = 3.5-4.0 \text{ MPa } \sqrt{\text{m}} \). This tendency becomes much more remarkable when crack closure is taken into account, i.e. FCP rates are characterized in terms of \( \Delta K_{\text{eff}} \) where the slope change occurs at \( \Delta K_{\text{eff}} = 2.5-3.0 \text{ MPa } \sqrt{\text{m}} \). Similar slope changes in FCP characteristics have also been reported in aluminium alloy [6] and pure titanium [7] and may be related to the transition from structure-sensitive to structure-insensitive behaviour.

Based on fracture surface analysis, it was confirmed that the transition of the operative micromechanisms of fracture has occurred. Fig. 4 reveals typical examples of SEM micrographs showing fracture surface appearance. Above the transition (Fig. 4(a)), fracture surface is totally covered with facets with many straight lines or steps whose directions are close to the macroscopic FCP direction. Each facet seems to correspond with grain. Below the transition (Fig. 4(c)), on the other hand, the predominant fracture mode is quasi-cleavage, which is definitely different from the fracture surface appearance above the transition. Each cleavage facet also seems to correspond with grain.

Near the transition (Fig. 4(b)), both fracture modes coexist.

The transition of fracture mechanism is believed to occur when the cyclic plastic zone size is equal to the average grain size. The cyclic plastic zone size, \( r_{\text{cp,eff}} \), for plane strain state is given by the following equation.

\[
r_{\text{cp,eff}} = \frac{1}{2\sqrt{2\pi}} \left( \frac{\Delta K_{\text{eff}}}{2\sigma_y} \right)^2
\]

where \( \sigma_y \) is the yield strength of material.

Substituting \( \Delta K_{\text{eff}} \approx 3.0 \text{ MPa } \sqrt{\text{m}} \) and \( \sigma_y = 110 \text{ MPa} \) into Eq. (1), the \( r_{\text{cp,eff}} \) value obtained is approximately 21 \( \mu \text{m} \) which is not in agreement with the average grain size of 60 \( \mu \text{m} \). The difference is rather large, but the scatter in grain size should be considered.

It is interesting to compare the FCP characteristic of the present Mg alloy with that of other light metals, i.e. two types of aluminium alloys, 7075-T6 [8] and 6063-T5 [9], and pure titanium [10]. The reasons why those materials were selected for comparison are that 7075-T6 alloy is very sensitive to environment, while 6063-T5 alloy has a good corrosion resistance, and pure titanium has the same crystallographic structure as Mg alloy and very excellent corrosion resistance. Fig. 5 represents the relationship between FCP rate and \( \Delta K_{\text{eff}} \) normalized by elastic modulus, \( E \), \( \Delta K_{\text{eff}}/E \). 6063-T5 alloy and pure titanium show nearly the same FCP behaviour, while 7075-T6 alloy exhibits faster FCP rate than those alloys in the entire \( \Delta K_{\text{eff}}/E \) region, particularly remarkable in high \( \Delta K_{\text{eff}}/E \) region. It should be noted that the Mg alloy has the worst FCP resistance. In high \( \Delta K_{\text{eff}}/E \) region, the Mg alloy shows the same FCP behaviour as 7075-T6 alloy, but significantly faster FCP rate in low \( \Delta K_{\text{eff}}/E \) region, i.e. below the transition. Therefore, it is believed that the observed differences in FCP behaviour would be attributed to corrosion resistance of the Mg alloy.

Mg alloys have very poor corrosion resistance and it has been indicated that air was aggressive environment [11] and corrosion fatigue took place above 80\% relative humidity [12]. Therefore, the quasi-cleavage fracture observed below the transition would be induced by moisture in laboratory air, because FCP rates are slow in that region and thus water would be condensed within the crack by the capillary action, leading to moisture-assisted FCP characteristic. In order to confirm this hypothesis, FCP tests in various corrosive environments are now in progress.

Detailed descriptions on the orientation and stress ratio dependence of FCP behaviour of the present Mg alloy are given elsewhere [13].
3.2. Fatigue strength

The $S$–$N$ diagram is shown in Fig. 6. It is well known that fatigue limit does not exist in non-ferrous alloys, thus fatigue tests were continued until $3 \times 10^7$ cycles. At $\sigma = 50$ MPa, fatigue failure did not occur, but non-propagating cracks were observed on the specimen surface. As can be seen in the figure, the present Mg alloy seems to possess a definite fatigue limit and similar behaviour is also recognized in extruded AZ61 alloy [12], which may be related to the existence of non-propagating cracks.

Two particular points can be established in the fatigue behaviour of the present alloy. One is the fatigue ratio ($\sigma/\sigma_y$) at $10^7$ cycles of 0.23, indicating that the alloy has considerably lower relative fatigue strength. Ogarrevic and Stephens have indicated that fatigue ratio was between 0.25 and 0.5 for wrought Mg alloys and higher ratios were for higher strength alloys [1]. The other is the fracture mode depending on applied stress level as shown in the figure. Above 70 MPa, cracks initiate at the surface (surface fracture), while below that stress at the interior of the specimen (subsurface fracture). Subsurface fracture is a very interesting finding, which has not been recognized in Mg alloys, as far as the authors know.

In the following sections, surface fracture and subsurface fracture will be discussed separately.

3.3. Surface fracture

3.3.1. Crack initiation

Crack initiation was examined in detail for the main cracks that led to final failure and many other cracks remained on the surface of failed specimens. Consequently, it was found that cracks have initiated within grains (transgranular) or at grain boundaries (intergranular). Initiation in both transgranular and intergranular modes occurred together in nearly equal amounts.

Fig. 7 reveals typical examples of SEM micrographs showing crack initiation. Fig. 7(a) is transgranular mode and intense slip bands can be seen parallel to the crack, thus the crack generates due to cyclic slip deformation. Fig. 7(b) is intergranular mode and intense slip
bands are also seen, which have a certain angle against the grain boundary at which the crack initiated. This may indicate that slip bands are blocked by the grain boundary and thus a stress concentration occurs, then leading to intergranular crack initiation.

3.3.2. Small crack growth

Small crack growth was measured at three different applied stresses of 70, 60 and 55 MPa. Surface crack length, $2c$, is represented in Fig. 8 as a function of cycle ratio, $N/N_f$ ($N_f$: fatigue life). As clearly seen in the figure, crack initiation occurs at an extremely early stage of fatigue life regardless of stress level. It should be noted that at $\sigma = 55$ MPa, the crack initiated at the interior of the specimen, but it appeared on the surface immediately, indicating that subsurface crack initiation also occurred at very early stage of fatigue life. In addition, at $\sigma = 70$ MPa, multiple cracks initiated, grew and then often coalesced each other. These findings imply that the crack initiation resistance of the present alloy would be significantly low, which is believed to be a cause for the low relative fatigue strength observed. Also suggested from the figure is the importance of the evaluation on small crack growth behaviour, because fatigue life is mostly dominated by small crack growth.

The relationship between crack growth rate, $da/dN$, and maximum stress intensity factor, $K_{\text{max}}$, for small cracks is shown in Fig. 9, where crack depth, $a$, and $K_{\text{max}}$ were obtained by assuming the aspect ratio of $a/c = 1$. Also included are the $da/dN - \Delta K$ and $da/dN - \Delta K_{\text{eff}}$ relationships for large cracks shown in Fig. 3. Small cracks exhibit the well-known behaviour in the present alloy, i.e. they can grow below the threshold effective stress intensity factor range, $\Delta K_{\text{eff,th}}$, for large cracks and show frequent fluctuations in FCP rate. An example of early small crack growth behaviour is demonstrated in Fig. 10, where surface crack growth rate at both crack tips, $dc/dN$, are represented as a function of crack tip location, i.e. crack length, from the crack initiation site, $c$. As can be seen in the figure,

![Image](image_url)
FCP rates fluctuate largely at the crack size of $2c < 0.4$ mm. The authors have indicated that FCP rates for small cracks were often decreased by microstructural barriers such as grain boundary in a wide variety of metals [14]. The present alloy seems to show more pronounced fluctuations compared with other materials.

The crack path profiles are shown in Fig. 11. Small deflection and branching can be seen, which was confirmed to take place at grain boundaries when the cracks grew into the neighbour grain. This may be due to limited slip systems of the present alloy that has a hexagonal close-packed crystal structure. The authors have reported that remarkable fluctuations in FCP rate occurred in pure titanium having the same crystal structure [15,16] and also indicated that the crack size below which fluctuations were seen (microstructurally small crack) was approximately eight times the grain size in a wide variety of metals [14]. The present alloy also conforms to such a relationship.

It is seen in Fig. 9 that the $da/dN - K_{\text{max}}$ relationships for small cracks nearly coincide with the $da/dN - \Delta K_{\text{eff}}$ relationship below the transition for large cracks, except for both the data at $\sigma = 70$ MPa and the early crack growth region. Fig. 12 reveals SEM micrographs showing fracture surfaces for small cracks. At both applied stresses, a stage I facet can be seen at the crack initiation site, thus the cracks initiated due to cyclic slip deformation, as indicated previously. Fracture surfaces after crack initiation are ductile and facets with steps can be seen, which is the same appearance as observed at high $\Delta K_{\text{eff}}$ region for large cracks (Fig. 4(a)). Such appearance is definitely different from that seen in low $\Delta K_{\text{eff}}$ region, i.e. below the transition. Therefore, it should be emphasized that although the $da/dN - K_{\text{max}}$ relationships for small cracks coincided with the $da/dN - \Delta K_{\text{eff}}$ relationship for large cracks, the operative fracture mechanisms are different between both cracks.

3.4. Subsurface fracture

3.4.1. Morphology of subsurface crack initiation

As seen in Fig. 6, subsurface crack initiation occurred at applied stresses below 70 MPa. Typical examples of SEM micrographs showing subsurface crack initiation site are shown in Fig. 13. A notable feature is the presence of facets at the crack initiation site regardless of stress level. Nonmetallic inclusions that have been often observed in high strength steels and surface-modified steels are not seen, thus the subsurface crack initiation mechanism is definitely different from those steels. The facets are basically flat and smooth, but one can recognize parallel straight marks on their surface (Fig. 13(b)).

The authors have observed subsurface fracture in beta titanium alloys and indicated the presence of smooth facets at the crack initiation site [17,18]. Alpha phase in titanium alloys has the same hexagonal close-packed crystal structure as Mg alloy, thus the
A comparison of the subsurface crack initiation between both materials seems to be significant. The morphology of the subsurface crack initiation site in beta Ti–22V–4Al alloy is revealed in Fig. 14 [18]. It is worth noting that the features of the facets are very similar to those observed in the present Mg alloy. Subsurface crack initiation in beta titanium alloys has not been fully understood, but it is believed that grain boundary alpha phase is related to crack initiation [19,20]. Also in alpha-beta titanium alloys, subsurface crack initiation has been recognized and it has been proposed that facets resulted in due to cleavage of alpha phase because of the restricted operative slip systems [21]. At the present time, subsurface crack initiation mechanism of the present Mg alloy is not known, but it is suggested that the facets would be formed due to crystallographic factor such as slip or cleavage, because of the same crystal structure as alpha phase in titanium alloys and the presence of parallel straight marks on the facet surface.

3.4.2. Features of subsurface crack initiation

At \( \sigma = 70 \) MPa, both surface and subsurface crack initiations were seen on the same fracture surface as shown in Fig. 15. Fig. 15(a) and (b) show surface and subsurface crack initiation, respectively. At the subsurface crack initiation site, a facet is present, which is similar to those shown in Fig. 13. Both cracks grew independently and then coalesced immediately before final failure. Below 70 MPa, surface crack initiation also occurred as indicated in Fig. 6, thus fracture mode does not change suddenly and there is a stress range in which both initiation modes coexist and with a further decrease in applied stress, only subsurface crack initiation would occur.

Fig. 16 represents the relationship between facet size, \( \sqrt{\text{area}} \), and distance from surface to facet, \( d_{\text{facet}} \). It can be seen that except for the facet at \( \sigma = 70 \) MPa, the facet sizes are nearly the same as, or smaller than, the average grain size. This would again suggest a crystallographic nature of the formation of the facets. Also seen in the figure are the \( d_{\text{facet}} \) values below 400 \( \mu \)m and thus subsurface cracks initiate at the location relatively close to the surface even under axial loading with no stress gradient.

Finally, the relationship between the maximum stress intensity factor for facet, \( K_{\text{max,i}} \) and \( N_f \), is shown in Fig. 17. As can be seen in the figure, there is no correlation between \( K_{\text{max,i}} \) and \( N_f \). On the contrary, in titanium alloys, a good correlation between both has been established [18]. Such difference between the present Mg alloy and titanium alloys would be attributed to subsurface crack initiation period.

![Fig. 14](image1.png)

Subsurface crack initiation site in solution-treated and aged materials of beta Ti–22V–4Al titanium alloy: (a) STA750; (b) STA800 [18].

![Fig. 15](image2.png)

Two different modes of crack initiation observed at \( \sigma = 70 \) MPa: (a) surface; (b) subsurface.

![Fig. 16](image3.png)

Relationship between facet size and distance from surface to facet.

![Fig. 17](image4.png)

Relationship between maximum stress intensity for facet and fatigue life.
4. Conclusions

Axial fatigue tests have been performed using smooth specimens and CT specimens of a rolled AZ31 Mg alloy in laboratory air at ambient temperature. Fatigue strength and fatigue crack propagation (FCP) characteristic were evaluated and fracture mechanism was discussed on the basis of crack initiation, small crack growth and fracture surface analysis. The conclusions can be made as follows:

1. The relationship between FCP rate and stress intensity factor range, \( \Delta K \), for large cracks consisted of two sections with different slopes, which became much more remarkable in the FCP behaviour after allowing for crack closure, i.e. characterized in terms of the effective stress intensity factor range, \( \Delta K_{\text{eff}} \). This was due to the transition of the operative micromechanisms of fracture.

2. The present Mg alloy had the worst FCP resistance for large cracks when compared with two types of aluminium alloys, 7075-T6 and 6063-T5, and pure titanium.

3. Fatigue strength at 10^7 cycles was 50 MPa that led to a considerably low fatigue ratio of 0.23.

4. Fracture mode changed around 70 MPa, above which surface fracture occurred, while below that stress subsurface fracture took place.

5. In surface fracture, cracks initiated at very early stage of fatigue life due to cyclic slip deformation in transgranular or intergranular mode and both initiation modes occurred together in nearly equal amounts.

6. Except for early growth immediately after crack initiation, small crack growth behaviour coincided with the crack propagation characteristic below the transition for large cracks, but the operative fracture mechanisms were different between small and large cracks.

7. In subsurface fracture, facets were always present at the crack initiation site, which were located mostly close to the surface, and the facet sizes were nearly the same as, or smaller than, the average grain size.

8. There existed no correlation between the maximum stress intensity factor for facet and fatigue life.

References