An experimental study of the orientation effect on fatigue crack propagation in rolled AZ31B magnesium alloy

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Abstract

Mode I fatigue crack growth (FCG) experiments were performed using compact tension (CT) specimens made of rolled AZ31B magnesium alloy in ambient laboratory air. The testing specimens were made with respect to two material orientations: a crack surface perpendicular to the rolled direction (R-T) and a crack perpendicular to the thickness or normal direction (N-T). The constant amplitude load experiments were performed at three load (R) ratios (minimum load over maximum load in a loading cycle) of 0.1, 0.5, and 0.75, respectively. Material orientation was found to affect the early crack growth stage more than the later stable growth stage. For each R ratio, the threshold stress intensity factor range for the R-T specimens was less than that for the N-T specimens. Three sub-stages of steady crack growth were observed following the threshold stage: a low Paris law slope for the first sub-stage, a second sub-stage with a very high slope, and an intermediate slope during the third sub-stage. The R-T specimens exhibited an overall typical Mode I cracking direction, with occasional local deviation from the horizontal crack path. The N-T specimens displayed a general Mode I cracking with irregular crack pathing into the specimen away from the observation surface. Transgranular cracking was the primary cracking mode for both specimen orientations. Slip induced cleavage dominated cracking in both orientations. Few residual twins were found in the plastic zone area surrounding the crack tip, and no evidence of twin boundary cracking was found.

Keywords: light microscopy, EBSD, magnesium alloys, fatigue crack growth, orientation relationships, R -ratio

1. Introduction

Magnesium (Mg) alloys could significantly impact the future of structural materials due to their excellent physical properties such as exceptional specific strength and good machinability. In particular, the automotive industry is attempting to utilize these properties to help improve the fuel economy of vehicles [1]. Over the past two decades, studies were conducted to explore, understand, and improve the mechanical properties of Mg and its alloys.

Mg and its alloys have a hexagonal close-packed (HCP) crystal structure, which mechanically behaves differently from the more commonly used cubic structured materials due to fewer and more difficult to activate slip systems. Mg and its alloys deform primarily through a combination of two mechanisms: $(0\ 0\ 0\ 1)$ basal slip and $\{1\ 0\ \bar{1}\ 2\}$ tensile twinning, the latter of which accommodates strain along the $\langle c \rangle$ direction of the HCP crystal [2]. These two

deformation mechanisms have a significant impact on crack initiation and propagation [3-8]. Additionally, the strong directionality of the deformation mechanisms results in a crack path that can deviate from the typical horizontal Mode I cracking direction seen in traditional cubic structures [8-13].

The deformation mechanisms and mechanical behavior in Mg alloys are dependent on the microstructure. Fatigue behavior is improved through grain refinement. In particular, reducing the grain size increases the stress required to activate twinning [14]. Barnett *et al* [15] performed compression tests on extruded Mg-3Al-1Zn to examine the deformation mechanisms as a result of varying grain sizes and temperatures. It was found that there was a transition of deformation mechanism, from twinning to dislocation slips, as the grain size was reduced. Horstemeyer *et al* [16] performed fully reversed fatigue tests on commercial high-pressure die cast automotive AZ91E-T4 Mg alloy. Their results showed that, for a given stress intensity factor range, the fatigue crack growth (FCG) rate was higher in a fine-grained microstructure than in a coarse-grained microstructure.

Metal processing can form a strong texture that affects the FCG behavior in wrought Mg alloys. Rolling or extrusion processes result in a majority of grains with their c-axes oriented in a particular direction relative to the working direction [9,17-23]. The texture formed due to the material processing leads to an anisotropic mechanical behavior. Twinning-detwinning during cyclic loading results in directional anisotropy and asymmetric stress-strain hysteresis loops under tension-compression loading [24-26]. Rolled Mg alloys typically contain a strong basal texture with the c-axes perpendicular to the rolled direction, along the “normal direction” [17,18]. Morita *et al* [18] performed tension-compression tests on hot-rolled AZ31B Mg alloy. They reported that the FCG rate of a crack propagating perpendicular to the rolled direction was lower than that of a crack propagating parallel to the rolled direction at a similar stress intensity factor range. Wu *et al.* [28] also performed cyclic loading tests with a load ratio or R-ratio (minimum load over maximum load in a loading cycle) of 0.1 on rolled AZ31B compact tension (CT) specimens to examine the twinning-detwinning behavior surrounding and extending from the crack tip. They used specimens of the N-R orientation, where the loading direction is along the normal direction, which is parallel to the c-axes of many grains. It was found that reversible twinning-detwinning is the dominant deformation mechanism for the material studied, yet also revealed that only a small amount of residual twins remained following the loading. Morita *et al* [29,30] performed FCG experiments on rolled AZ31B specimens at a R-ratio of 0.1 to examine the fracture surfaces. They found that crack propagation along the a-axis direction is favorable. They also reported very few residual twins along the crack path of the R-T orientation. Alternatively, extrusion results in a typical texture where the basal planes are primarily oriented parallel to the extrusion direction [9,21-23]. It was found that the crack growth resistance in the direction perpendicular to the extrusion direction is greater than that in the direction parallel to the extrusion direction. This effect is attributed to the basal texture and lamellar microstructure along the extrusion direction [9,19,20].

Loading parameters also affect the FCG behavior. An increase in R-ratio has been shown to increase the FCG rate in Mg alloys [9,29,32]. Zeng *et al.* [11,32,33] performed FCG experiments on single-edge notched plate specimens prepared from AZ61 and AZ80 Mg alloys. It was reported that, for a given stress intensity factor range, increasing the loading frequency resulted in a decrease of FCG rate within the studied alloy. Zheng *et al* [34] examined the effects of overload and two-step loading on extruded AZ31B. A single tensile overload was found to greatly reduce the FCG rate a short length prior to returning to the expected FCG rate. The

overload influencing zone was smaller in the extruded AZ31B than in other metallic materials [35-37]. Their results from the two-step loading experiments revealed that maintaining a constant maximum load in a loading cycle when reducing the R-ratio, mean load, or other loading parameter results in less retardation of the FCG rate.

While research on the FCG behavior of Mg and its alloys has increased in the past two decades, their behavior is not as well documented as other traditional metallic materials. In particular, the effect of specimen orientation and texture on FCG behavior has not been thoroughly examined. In the present study, the fatigue crack propagation behavior of rolled AZ31B Mg alloy is studied. The effects of specimen orientation in relation to the rolled direction and R-ratio on the FCG behavior are considered. Two specimen orientations and three different R-ratios are used.

2. Experiments

2.1 Material and microstructure

The material used in the current study is a commercially acquired rolled plate of AZ31B Mg alloy (Mg-3Al-1Zn). No heat treatment was applied after rolling. Fig. 1a is a three-dimensional isometric representation of the microstructure and initial texture of the rolled AZ31B Mg alloy obtained by electron backscatter diffraction (EBSD). The three planes have homogeneous microstructures consisting of mostly equiaxed grains with an average diameter/size of approximately $43\mu\text{m}$. Scattered residual twins as a result of rolling were found in the initial material, but not visualized in Fig. 1. Based on Fig. 1, this material shows a strong basal texture in which the c-axes of most grains align perpendicular to the R-T plane along the normal direction.

Fig. 1. Initial microstructure and macro texture of rolled AZ31B Mg alloy

2.2 Specimen

Rectangular compact tension (CT) specimens were prepared for two orientations as shown in Fig. 2 where “RD,” “ND,” and “TD” denote the rolled/longitudinal, normal, and transverse directions, respectively. A designation of X-Y was used to identify the CT specimens.

For Mode I crack growth, “X” designates the direction of the applied load and “Y” designates the cracking direction. For example, an R-T specimen is loaded in the rolled/longitudinal direction and the crack propagates along the transverse direction. For the other orientation, an N-T specimen is loaded in the normal direction and the crack grows in the transverse direction.

Fig. 2. Compact tension (CT) specimens of rolled AZ31B Mg alloy and two specimen orientations with respect to the rolled plate

A sharp notch was made on the specimen using electrical discharge machining (EDM). Prior to testing, one specimen surface of each specimen was polished to a mirror-like finish by standard metallographic preparation techniques to facilitate crack observation and measurement. The specimen surfaces were sequentially ground using 360/P500 to 1000/P3200 silicon carbide grit papers. Polishing was done using 6 μm , followed by 1 μm oil based polycrystalline diamond suspensions. A final polish was achieved using 0.05 μm MasterPolish® suspension. The specimens were then etched in an Acetic-Picral solution (10 mL acetic acid, 4.2 g picric acid, 10 mL distilled water, and 70 mL ethanol) until surface browns, which usually occurred after approximately five seconds, to reveal the initial microstructure. Companion EBSD specimens were prepared in a similar manner with only being etched in 3% natal for 10 seconds to remove the final surface layer of damage.

2.3 Experiments

A servo-hydraulic load frame with a ± 25 kN load capacity was used for testing the CT specimens. In order to accommodate the small load used in the experiments, a 2.0 kN load cell was used for the control of the load. All the experiments were conducted with constant amplitude load employing a sinusoidal waveform. The loading frequency was varied as a function of stress intensity factor range (ΔK), where the loading frequency was gradually decreased with increasing ΔK . A frequency of 20 Hz was used near threshold and the frequency was decreased to 2 Hz near the fast fracture region. Three load or R ratios (minimum load over maximum load in a loading cycle) were used for the two specimen orientations. These values were 0.1, 0.5, and 0.75. The threshold stress intensity factor range, ΔK_{th} , was determined by sequentially reducing the load amplitude for a given R -ratio until the FCG rate was less than 1.0

$\times 10^{-7}$ mm/cycle. When the R-ratio was 0.5 and 0.75, the crack was initiated with an R-ratio of 0.1, and sequentially increased to the desired R-ratio with a step size of 0.05 while maintaining a constant maximum load to reduce the overload effect. The load amplitude was then sequentially reduced until threshold value was obtained. After the determination of the threshold value, the FCG experiment was conducted by retaining the same R-ratio used to find ΔK_{th} and adjusting the load amplitude so that the initial stress intensity factor range was above the threshold value.

The stress intensity factor range of a rectangular CT specimen was calculated using the following equation [38]:

$$\Delta K = \frac{\Delta P}{B\sqrt{W}} \frac{(2+\alpha)}{(1-\alpha)^{1.5}} (0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4) \quad (1)$$

where ΔP was the applied load range, B was the specimen thickness, W was the specimen width, and α was the crack length over specimen width, or a/W . The crack length was measured from the line of action of the applied load to the crack tip and the specimen width, W , was measured from the line of load application to the far edge of the specimen, as shown in Fig. 1. The crack length was measured using a traveling long distance microscope QM100 with an image capture resolution of 1 μ m. In the cases where $\alpha < 0.2$, an adjustment based on Fig. 3 was made according a finite element analysis performed on a CT specimen [35].

Fig. 3. Stress intensity factor range correction curve

EBSD was performed using a JEOL 7100F field emission scanning electron microscope (SEM) with an Oxford Instruments Nordlys-f model EBSD camera. An accelerating voltage of 25 kV, probe current of 10, and an objective lens 3 were used. The specimen was tilted to 70° in the chamber. The working distance varied by specimen but was within the range of 15 mm to 20 mm. The resulting inverse pole figure (IPF) maps were cleaned using wild spike extrapolation and 7-neighbor correction methods within the Oxford HKL Channel 5 software.

Mode I loading of an isotropic material usually results in a crack path perpendicular to the loading direction. However, for strong textured materials, the crack path may deviate. For convenience of discussion, “horizontal direction” will be used in reference to the direction

perpendicular to the loading action, which is the usual Mode I cracking direction for a homogeneous material. In the cases where the crack path deviates from the horizontal direction, the crack length measured will be the projection of the actual crack position onto the horizontal direction.

Table 1 summarizes the FCG experiments conducted where a_0 is the initial crack length measured from the line of action of the applied load to the crack tip when the experiment was initiated, a_{fin} is the final crack length when the experiment was concluded, and ΔK_{fin} is the final stress intensity factor range when the experiment was terminated. To determine the FCG rate, da/dN , a parabolic curve is applied to best fit five consecutive points corresponding to the relationship between the crack length and the loading cycle number. The derivative of the fitted parabolic curve is applied to the middle loading cycle of the five points used, and the resulting value is reported as the experimental FCG rate. The fatigue crack paths are characterized using light optical microscopy (OM) following the FCG experiments.

Table 1. Summary of fatigue crack growth experiments for rolled AZ31B Mg alloy.

Specimen	R	P_{max} (kN)	a_0 (mm)	a_{fin} (mm)	ΔK_{th} (MPa $\sqrt{m}$)	ΔK_{fin} (MPa $\sqrt{m}$)
N-T07	0.1	0.211	5.31	17.27	1.98	9.21
N-T10	0.1	0.333	3.89	14.60	N/A	9.02
N-T14	0.1	0.278	2.65	14.22	N/A	7.31
N-T18	0.1	0.264	2.53	16.06	2.00	9.65
N-T05	0.5	0.260	6.50	13.45	1.54	3.69
N-T16	0.5	0.266	4.21	13.30	1.36	3.64
N-T04	0.75	0.344	5.37	12.45	0.93	2.14
N-T19	0.75	0.288	8.59	16.14	1.07	2.91
R-T03	0.1	0.111	10.98	24.25	1.23	9.86
R-T08	0.1	0.333	4.44	17.70	1.49	9.71
R-T09	0.1	0.400	4.63	10.19	N/A	8.67
R-T13	0.1	0.289	6.87	18.93	1.56	8.67
R-T25	0.1	0.229	1.84	20.43	1.47	9.14
R-T07	0.5	0.200	5.93	18.99	0.85	3.46
R-T10	0.75	0.336	5.77	17.36	0.77	2.36

Note: R, load ratio; P_{max} , maximum load in a loading cycle; a_0 , crack length prior to the constant amplitude loading; a_{fin} , crack length measured when the experiment was concluded; ΔK_{th} , threshold stress intensity factor range measured for a particular specimen; ΔK_{fin} , stress intensity factor range when the experiment was concluded

3. Results

3.1 Crack growth in two specimen orientations

The relationship between the FCG rate and the crack length, as well as the stitched micrograph of the overall crack path, for an R-T specimen with a R-ratio of 0.1 (R-T25) can be seen in Fig. 4. High-magnified micrographs highlighting notable features are also shown. The results reveal that the FCG rate increases monotonically with the crack extension under constant amplitude loading. However, there is a nearly instantaneous increase in crack growth rate

around $a=13.3\text{mm}$, corresponding to a ΔK value of $3.8\text{ MPa}\sqrt{\text{m}}$, for this specimen. After this transition, residual deformation bands, either twins or slip bands, are present along the crack path, and increase in number with increasing crack length. This is evident by the progression seen along the crack path. Fig. 4a, pertaining to a ΔK value of approximately $2.8\text{ MPa}\sqrt{\text{m}}$, reveals that there are no deformation bands present along the crack path prior to the transition. The progression of deformation bands observed from Fig. 4b to 4c could indicate that the bands are visible following the transition and increase in quantity as the crack length and FCG rate increase. However, the exact nature of these bands cannot be determined by optical microscopy alone.

Fig. 4: Crack growth rate as a function of crack length, stitched OM micrographs revealing the crack path, and OM micrographs captured at different positions along the crack path for specimen R-T25.

The stitched micrograph in Fig. 4 reveals that the crack follows an overall Mode I direction in a transgranular mode. Local crack deviation from the horizontal direction was noticed, but no explicit patterns were identified. For this particular specimen, the crack path remained consistently along the horizontal, only showing a slight downward incline of 4° below the horizontal.

For the R-T specimen orientation, secondary cracking is present but infrequent and is difficult to identify as the secondary cracks only grow a short distance and usually re-connect with the primary crack. The section of material between the two cracks is often lost during the experiment, especially if the experiment is carried to a large ΔK . Examples of this effect can be seen in micrographs B and C in Fig. 4 where the crack edges are not well aligned.

Fig. 5: Crack growth rate as a function of crack length, stitched OM micrographs revealing the crack path, and OM micrographs captured at different positions along the crack path for specimen N-T14.

The relationship between FCG rate and crack length, as well as the stitched micrograph of the overall crack path, for an N-T specimen with a load ratio of 0.1 (N-T14) can be seen in Fig. 5. High-magnification micrographs highlighting notable features are also shown. This specimen orientation follows one of two possible FCG features. One is a general monotonic increase in crack growth rate as a function of crack length as a result of Mode I cracking direction. The second is where the crack paths into the specimen, unobservable through the long distance microscope. The crack then reappears at a similar horizontal distance from the notch, but is some distance below the initial crack tip. The crack continues to propagate from this new location along the Mode I cracking direction. As seen in the stitched micrograph in Fig. 5, the initial crack follows the typical Mode I cracking direction for approximately 3.0 mm before

appearing to arrest. The FCG rate is reduced until the crack reappears under the microscope. This is evident by the sudden drop in FCG rate when the crack reaches about 3.0 mm, followed by a plateau of growth for the following 1.0 mm. It is important to note that the crack appears to propagate both towards the notch and in the Mode I direction after it reemerges. Eventually, the portion progressing toward the notch connects with the initial crack path, appearing as if there had been a crack branch when it was actually a single crack. This irregular crack path results in the erratic FCG rate observed early in the experiment, and does not occur after the crack reaches a length of approximately 9 mm, or a value of $4.6 \text{ MPa}\sqrt{\text{m}}$ for ΔK .

Accounting for the position where the crack reemerges in this specimen, Mode I cracking behavior was observed in a transgranular method. The locations where the crack strongly deviates from the horizontal path are due to a similar behavior as described in the previous paragraph. Examples of this behavior can be seen in Fig. 5a and 5b. Fig. 5a shows the result of the interior crack path similar to what is described in the previous paragraph. Fig. 5b shows the same behavior as Fig. 5a, but shows only one connection between the initial crack path and the newly revealed crack location. This type of connection may be falsely identified as a nearly vertical crack. The crack portion going in the reverse direction from the normal crack direction also gives a false image of the crack propagating toward the notch. Such cracking behavior is a result of the crack inside the specimen opening to the surface. Aside from these connecting cracks, secondary cracks were infrequently found.

Similar to what is observed in the R-T orientation specimen, deformation bands were present in the N-T orientation specimen. However, there is no noticeable pattern in regards to their initiation point relative to the crack length or FCG rate. There are a few noticeable bands present as early along the crack path as what is seen in Fig. 5a. However, due to the location of this micrograph relative to the overall crack path, it may be that the bands present in this area are a result of a potential increase in local stress due to the connecting cracks. This may especially be true due to the lack of deformation bands along the initial crack and along the primary crack path between the two intersections. Deformation bands similar to what was seen in the R-T orientation do not appear until shortly after Fig. 5b when the crack reaches a length of 7.7 mm corresponding to a ΔK value of $3.55 \text{ MPa}\sqrt{\text{m}}$. The amount of bands increases from that point until the experiment was terminated at a crack length of 14 mm and a ΔK of $7.3 \text{ MPa}\sqrt{\text{m}}$. Both specimen orientations appear to have a similar transition after which deformation bands become visible. The ΔK value for the N-T orientation of about $3.55 \text{ MPa}\sqrt{\text{m}}$, which is slightly less than the transition value of $3.8 \text{ MPa}\sqrt{\text{m}}$ found for the R-T orientation.

3.2 R-ratio and specimen orientation effects on FCG behavior

Fig. 6 summarizes the relationship between the FCG rate, da/dN , and the stress intensity factor range, ΔK , for the two specimen orientations at the three R-ratios studied. The arrows in the figure denote the threshold stress intensity factor range, ΔK_{th} . Certain tendencies are common for both material orientations at all three R-ratios. For a given ΔK value, an increased R-ratio results in an increased FCG rate, signifying the same general R-ratio effect observed in conventional metals. For both specimen orientations, increasing the R-ratio results in decreased threshold stress intensity factor range. The overall R-ratio effect was more significant in the N-T orientation than in the R-T orientation. For the N-T specimens, the threshold value ranges between $0.93 - 1.07 \text{ MPa}\sqrt{\text{m}}$ for a R-ratio of 0.75 and $1.98 - 2.01 \text{ MPa}\sqrt{\text{m}}$ for a R-ratio of 0.1.

When the R-ratio is 0.5, the stress intensity factor range was found to be between 1.3 and 1.6 MPa√m. The overall difference between the smallest and largest value is approximately 1.08 MPa√m. For the R-T specimens, the threshold value ranges between 0.756 – 0.768 MPa√m for a R-ratio of 0.75 and 1.23 - 1.564 MPa√m for a load ratio of 0.1. When the R-ratio is 0.5, the stress intensity factor range was found to be 0.825 MPa√m. The overall difference between the smallest and largest value is 0.8 MPa√m. For comparison, a threshold stress intensity factor range for a rolled AZ31 Mg alloy in the R-T orientation was found to be 1.90 MPa√m for a R-ratio of 0.05 [12], which supports the trend of increasing threshold stress intensity factor range with decreasing R-ratio. While not a large difference, the effect of R -ratio on ΔK_{th} was more significant on the N-T orientation than R-T orientation, especially at lower R -ratios. At larger values of ΔK , R -ratio limits the FCG rate as the experiment approaches the Mode I critical stress intensity factor, K_{IC} . As a result, larger FCG rates are achievable with a lower R -ratio.

Fig. 6. R-ratio effect on crack growth rate curves for both N-T and R-T specimens

Specimen orientation has a significant effect on the early crack growth regime for a given R-ratio. With the R-T orientation having a lower threshold stress intensity factor range than the N-T orientation, the steady crack growth regime in these specimens begins at a lower ΔK . However, when the R-ratio is 0.5 or 0.75, the FCG rate curves of the two orientations eventually converge and then coincide with each other, revealing that specimen orientation has

little effect at larger ΔK values. However, when the R-ratio is 0.1, this overlapping behavior is not seen. The FCG curve for the N-T orientation converges on the R-T orientation curve at a ΔK value of 3.5 MPa $\sqrt{m}$, but the curves do not coincide after they converge. This indicates that the influence of specimen orientation on crack growth behavior diminishes as the R-ratio increases.

Fig. 7. Crack growth rate curves for R-T specimens at a load ratio of 0.1

Fig. 8. Crack growth rate curves for N-T specimens at a load ratio of 0.1

Fig. 9. Crack growth rate curves for both specimen orientations at a load ratio of 0.5

Fig. 10. Crack growth rate curves for both specimen orientations at a load ratio of 0.75

A more detailed observation of the experimentally obtained crack growth results reveals unique features of the Mg alloy. To facilitate discussion, the results are re-presented in Figs.7-10 with each figure for one R -ratio. The Paris Law will be used for discussion,

$$\frac{da}{dN} = C(\Delta K)^m \quad (2)$$

where C is a material constant and m represent the slope of the crack growth da/dN - ΔK curve in log-log scale.

Rolled AZ31B Mg alloy displays three sub-stages characterized with three distinguishable Paris law slopes during stable crack growth as opposed to one single slope of the traditional materials. This behavior is most evident in the R-T orientation when the R -ratio is 0.1. Referring to Fig.7, a threshold stage similar to that of a conventional material is evident. In addition, an unstable crack growth stage is expected but not present in Fig.7 due to early termination of the experiments before final fracture. The crack growth between the threshold stage and the unstable crack growth stage is often referred to as the stable crack growth stage. Results shown in Fig.7 reveals a sigmoidal shaped crack growth curve, suggesting three sub-stages during stable crack growth. Sub-stage I is characterized with a small slope m . Sub-stage II has a very large slope m , signifying a rapid increase of FCG rate with an increase in stress intensity factor. The third sub-stage has a slope that is significantly smaller than that of the previous sub-stage but larger than that of sub-stage I.

With an R -ratio higher than 0.1, not all the three sub-stages are present. As shown in Figs. 9-10, part of sub-stage III was observed for $R=0.5$ and only sub-stages I and II were observed for $R=0.75$. For $R=0.5$ (Fig.9) and $R=0.75$ (Fig.10), the crack growth slope, m , in each sub-stage is

very similar to that of $R=0.1$ of the R-T orientation during the same sub-stage. Sub-stage II starts at an approximately crack growth rate of 2×10^{-5} mm/cycle and ends at 1.2×10^{-4} mm/cycle.

An exception to the three sub-stage stable crack growth in rolled AZ31B is the case for the N-T specimen at an R-ratio of 0.1 as can be seen in Fig.8. While the result of an individual testing specimen may show an irregular crack growth behavior, the overall tendency of the results from six testing specimens in the N-T orientation at $R=0.1$ can be best described by a constant Paris law slope, m , with a value of 3.8. Local deviations from this slope at lower ΔK values are a result of the irregular crack path shown in Fig. 5. After the ΔK value reaches about $4.1 \text{ MPa}\sqrt{\text{m}}$, the data fits very well with the Paris Law slope. Clearly, the material orientation and the R-ratio affect the stable crack growth.

Deviations from the traditional single-stage steady crack growth regime has been reported for other Mg alloys in literature. Zheng et al [9,34] performed similar experiments on three orientations of extruded AZ31B at three R-ratios of 0.1, 0.5, and 0.75. Their results found three sub-stages of steady crack growth in the L-T orientation at all three R-ratios, and sub-stage I was found to have a near zero m plateau. For the T-L orientation, a three sub-stage steady growth regime was seen when the R-ratio was 0.1 and 0.5. When the load ratio was 0.75, a single steady growth regime with a single slope was found. For the T-R orientation, a single steady crack growth regime of a single slope was found for all three load ratios. Uematsu et al [39] reported a two sub-stage steady growth regime for extruded AZ61 Mg alloy tested in the L-T material orientation with an R-ratio of 0.05.

4. Discussion

The fatigue crack growth experiments were performed on specimens designed for Mode I crack growth. Local crack deviation was observed in both specimen orientations and was more prevalent in the N-T specimens than in the R-T specimens. The crack growth displays an overall Mode I direction and therefore, Mode I stress intensity factor range was used for its convenience in discussion.

Fig. 11. Walker's model for R-ratio effect of rolled AZ31B

Recognizing that the Paris Law slopes were similar in the three R -ratios used, an attempt was made to use the following Walker's equation [37] to correlate the experiments with different R -ratios,

$$\Delta K_{eff} = K_{max}(1 - R)^{\nu} \quad (3)$$

In Eq.(3), ΔK_{eff} is the effective stress intensity factor range, K_{max} is the maximum stress intensity factor, R is the load ratio, and ν is a material constant. With a value of $\nu=0.5$, an overlap of the FCG data can be achieved. Using Walker's equation, Fig. 11 reveals more details about the FCG behavior of the two specimen orientations at the three R -ratios. The results show an overall division of three sub-stages for the stable crack growth with similar Paris law slopes to these discussed in the previous section. While not overlapping, the N-T orientation has a noticeably larger effective threshold stress intensity factor range than the R-T orientation. As a result of this, the range of sub-stage I in terms of ΔK is larger in the R-T orientation than in the N-T orientation. The overlap at sub-stage II reveals that this sub-stage is primarily dependent on specimen orientation and that R -ratio has negligible effect in this region. It is also revealed that the slope of sub-stage II is slightly larger in the R-T specimens than in the N-T specimens. As a

result, the increase of FCG rate is larger in the R-T specimens than in the N-T specimens as a result of sub-stage II. The culmination of these two factors is that the N-T orientation begins sub-stage III at both a lower ΔK_{eff} and a lower FCG rate than the R-T orientation. Similar to sub-stage II, sub-stage III is primarily dependent on specimen orientation due to the significant overlap within specimen orientation.

Compared to the results from similar experiments performed on extruded AZ31B CT specimens [9], some comparisons can be made. The microstructure following the extrusion process is different from the microstructure obtained from rolling. Large, elongated grains are found parallel to the extrusion direction, forming a lamellar microstructure of alternating elongated grains and small grain clusters. The plane perpendicular to the extrusion direction has equiaxed grains located in large and small grain clusters. Additionally, extrusion results in a strong basal texture parallel to the extrusion direction. As a result of different microstructure and texture, the cracking behavior for the extruded AZ31B specimens can vary from that of the rolled AZ31B specimens. One significant difference is that the extruded specimens did not display an explicit three sub-stage stable crack growth behavior in all orientations and R-ratios. For R-ratios of 0.1 and 0.5 in the T-L orientation, the three sub-stage stable crack growth regime can be seen. In particular, the extruded T-L curve with a R-ratio of 0.1 shows a similarity to that of the rolled R-T curve with a R-ratio of 0.1. Referring to the specimen orientation and material texture, it is noted that these two orientations share a similar texture. However, there is a difference of microstructure between the two specimen orientations, or more specifically, a difference in grain structure. This is the likely cause of the minor differences. The reported ΔK_{th} for the extruded T-L specimen was $\sim 1.5 \text{ MPa}\sqrt{\text{m}}$ when $R=0.1$. This value corresponds very well with the threshold values reported for the R-T specimens tested with a R-ratio of 0.1. As the R-ratio is increased, the two orientations become less similar. The ΔK_{th} for R-ratios of 0.5 and 0.75 were reported to be $\sim 1.5 \text{ MPa}\sqrt{\text{m}}$ and $\sim 1.1 \text{ MPa}\sqrt{\text{m}}$, respectively, for the extruded T-L orientation. The rolled R-T orientation had, for the same R-ratios, ΔK_{th} values of $0.83 \text{ MPa}\sqrt{\text{m}}$ and $0.77 \text{ MPa}\sqrt{\text{m}}$, respectively. Assuming that the two specimen types have the same texture and all else is consistent, microstructure is the primary factor affecting the threshold region at higher load ratios. Conversely, microstructure has an insignificant effect on the crack growth behavior at a lower R-ratio, which in this case is 0.1. There is no comparative texture to the rolled N-T orientation in the extruded specimens examined by Zheng et al [9].

Similar to the results found for the rolled AZ31B specimens, the predominant cracking mode in the extruded specimens was transgranular. However, intergranular cracking was found at the boundaries of extremely small grains within the small grain clusters in extruded AZ31B. The relatively large equiaxed grains in the rolled AZ31B specimens could explain why there was no observable intergranular cracking in the rolled AZ31B experiments.

Extra experiments were performed on EBSD companion specimens to examine the crystallography along the crack path and surrounding the crack tip. One companion specimen was prepared for each orientation and the experiments were performed at a R-ratio of 0.1 and load amplitude of 0.12kN. The experiments were concluded when ΔK reached $7 \text{ MPa}\sqrt{\text{m}}$. EBSD was performed along the crack path at ΔK values of $3 \text{ MPa}\sqrt{\text{m}}$ and $5 \text{ MPa}\sqrt{\text{m}}$, as well as at the crack tip when ΔK was $7 \text{ MPa}\sqrt{\text{m}}$. The crack path under EBSD reveals that the crack propagates in a transgranular mode for both orientations, and that residual twins can be located along the crack path and within the plastic zone surrounding the crack tip. Visually, more twins are identified in the N-T orientation than in the R-T orientation. For both orientations, the number of residual twins surrounding the crack increases as ΔK increases. While these twins are visible,

they are not as numerous as the deformation bands present in Fig. 4 and Fig. 5, micrographs B and C. Additionally, the crack does follow any twin boundaries. These observations indicate that most of the deformation bands present at higher ΔK values are slip bands. Using the general crystallography of the grains through which the crack is propagating, cracking mechanisms can be proposed.

Fig. 12: SEM-EBSD of N-T15 specimen along the crack path at ΔK of (a) 3 $\text{MPa}\sqrt{\text{m}}$, (b) 5 $\text{MPa}\sqrt{\text{m}}$, and at the crack tip at ΔK of (c) 7 $\text{MPa}\sqrt{\text{m}}$.

Fig. 13: SEM-EBSD of R-T15 specimen along the crack path at ΔK of (a) 3 MPa $\sqrt{\text{m}}$, (b) 5 MPa $\sqrt{\text{m}}$, and at the crack tip at ΔK of (c) 7 MPa $\sqrt{\text{m}}$.

Based on the texture shown in Fig. 1, two idealized crystal orientations are assumed for both orientations. For the N-T orientation, both idealized crystals have the loading direction along the c-axes of the crystals. The tensile loading direction along the c-axes of the hexagonal crystal in the N-T orientation is likely to produce $\{1\ 0\ \bar{1}\ 2\}$ -type twinning. Despite tension-tension loading, the plastic zone near the crack tip is subjected to cyclic tension-compression loading [28,41]. This results in twinning-detwinning within the plastic zone near the crack tip [28]. The small amount of residual twins along the crack path and within the plastic zone indicate that the majority of the twinned grains are fully detwinned upon unloading. The process of detwinning upon unloading is supported by findings from by Dong et al [42]. The detwinning process on unloading explains why there are few residual twins seen in Fig. 12. The crack path can be explained based on the ideal crystal orientations. In the N-T orientation, one crystal has the crack initiating perpendicular to the prismatic $(1\ 0\ \bar{1}\ 0)$ plane while the second has the crack initiating on an edge of the hexagonal structure, or perpendicular to the $(2\ \bar{1}\bar{1}\ 0)$ plane. Crack propagation along the basal slip plane would result in a horizontal, Mode I cracking direction in both ideal crystals. Due to the slight rotations of these ideal crystals within the real material, the majority of the deformation bands present in the OM micrographs for the N-T specimen are basal slip bands. With the crack propagating along the slip planes, twinning-detwinning is hindering the crack growth. This explains the decreased FCG rate in the N-T orientation for a given ΔK value relative to the FCG rate in the R-T orientation. The cause of the irregular crack path to the interior of the N-T specimen is unknown. A detailed examination of the fracture surface may reveal the driving force of the irregular crack path.

A similar idealized crystal model can be applied to the R-T orientation. For this orientation, a Mode I crack will initiate along $(2\ \bar{1}\ \bar{1}\ 0)$ $[1\ 0\ \bar{1}\ 0]$ or $(1\ 0\ \bar{1}\ 0)$ $[2\ \bar{1}\ \bar{1}\ 0]$. Unlike in the N-T orientation, the grain orientation relative to the loading direction is unlikely to produce twinning under tension. As a result, dislocation slips should be the dominant deformation mechanism. However, there is evidence of twinning at the crack tip when ΔK is at $7\ \text{MPa}\sqrt{\text{m}}$, indicating that the compressive local stress at the crack tip surpasses the twin activation stress around that value of ΔK . The lack of twinning at lower ΔK values in this orientation may explain the difference in FCG rate between the two orientations. The presence of twins along the crack path in the R-T orientation was also reported by Morita et al [29].

For the slip systems within the R-T orientation, basal slip is expected to be the primary slip mode, however, previous work on extruded AZ31B by Zheng et al [9] and on Mg single crystal by Ando et al [43-45] also reveals that $\langle c+a \rangle$ pyramidal slip may also be present. Examining the fracture surface, including EBSD, and exploring this crack path phenomena in both orientations may reveal the driving mechanisms for crack propagation in rolled AZ31B.

5. Conclusions

The following conclusions can be drawn from the results of the experimental study of crack growth behavior in rolled AZ31B magnesium alloy:

1. The stable crack growth in the traditional $\frac{da}{dN}$ - ΔK plot has three distinguishable sub stages, described as sub-stage I, II, and III, characterized by different Paris law slopes.
2. Load (R) ratio has a significant effect on the fatigue crack growth (FCG) behavior. The threshold stress intensity factor range decreases with increasing R-ratio for both specimen orientations. The effect is more significant in the N-T specimens than in the R-T specimens. The range of sub-stage I and sub-stage III FCG behavior significantly decreases with increasing R-ratio, with the presence of sub-stage III being negligible when the R-ratio was 0.75. Walker's equation can reasonably correlate the R-ratio effect on stable crack growth of the Mg alloy.
3. Specimen orientation has a significant influence on the crack growth, especially in the threshold stage I and when the R-ratio is 0.1. The N-T orientation has a significantly higher ΔK_{th} than that of the R-T orientation with the same R-ratio. The influence of the specimen orientation is marginal in the stable crack growth region when the R-ratio is higher than 0.1.
4. Transgranular cracking is the primary cracking mode in both specimen orientations. Slip induced cleavage dominates in both orientations.
5. Slip bands are observable after a distinguishable transition in the R-T orientation at an R-ratio of 0.1. The transition results in a rapid increase of FCG rate and occurs at a stress intensity factor range around $3.8\ \text{MPa}\sqrt{\text{m}}$.

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References

- [1] Mordike BL, Ebert T. Magnesium Properties – applications – potential. *Mat Sci Eng A* 2001; 302: 37-45
- [2] Yoo MH. Slip, twinning, and fracture in hexagonal close-packed metals. *Metall Mater Trans A* 1981; 12(3): 409-18
- [3] Agnew SR, Yoo MH, Tomé CN. Application of texture simulation to understanding mechanical behavior of Mg and solid solution alloys containing Li or Y. *Acta Mater* 2001; 49(20): 4277-89
- [4] Koike J, Fujiyama N, Ando D, Sutou Y. Roles of deformation twinning and dislocation slip in the fatigue failure mechanism of AZ31 Mg alloys. *Scripta Mater* 2010; 63(7): 747-50
- [5] Yu Q, Zhang J, Jiang Y. Fatigue damage development in pure polycrystalline magnesium under cyclic tension-compression loading. *Mater Sci Eng A* 2011; 528(25-26): 7816-26
- [6] Yang F, Yin SM, Li SX, Zhang ZF. Crack initiation mechanism of extruded AZ31 magnesium alloy in the very high cycle fatigue regime. *Mater Sci Eng A* 2008; 491: 131–6.
- [7] Yu Q, Zhang J, Jiang Y. Direct observation of twinning-detwinning-retwinning on magnesium single crystal subjected to strain-controlled cyclic tension-compression in $[0\ 0\ 0\ 1]$ direction. *Philos Mag Lett* 2011; 91(12): 757-65
- [8] Yin SM, Yang F, Yang XM, Wu SD, Li SX, Li GY. The role of twinning-detwinning on fatigue fracture morphology of Mg – 3%Al – 1%Zn alloy. *Mater Sci Eng A* 2008; 494(1-2): 397-400
- [9] Zheng S, Yu Q, Jiang Y. An experimental study of fatigue crack propagation in extruded AZ31B magnesium alloy. *Int J Fatigue* 2013; 47: 174-83
- [10] Nicoletto G, Konecna R, Pirondi A. Fatigue crack paths in coarse-grained magnesium. *Fatigue Fract Eng Mater Struct* 2005; 28: 237–44.\
- [11] Zeng RC, Han EH, Ke W. A critical discussion on influence of loading frequency on fatigue crack propagation behavior for extruded Mg–Al–Zn alloys. *Int J Fatigue* 2012; 36: 40–6.
- [12] Tokaji K, Kamakura M, Ishiizumi Y, Hasegawa N. Fatigue behavior and fracture mechanism of a rolled AZ31 magnesium alloy. *Int J Fatigue* 2004; 26: 1217-24
- [13] Venkateswaran P, Ganesh Sundara Raman S, Pathak SD, Miyashita Y, Mutoh Y. Fatigue crack growth behavior of die-cast magnesium alloy AZ91D. *Mater Letters* 2004; 58: 2525-2529
- [14] Meyers MA, Vöhringer O, Lubarda VA. The onset of twinning in metals: a constitutive description. *Acta Mater* 2001; 49(19): 4025-39
- [15] Barnett MR, Keshavarz Z, Beer AG, Atwell D. Influence of grain size on the compressive deformation of wrought Mg-3Al-1Zn. *Acta Mater* 2004; 52: 5093-103
- [16] Horstemeyer MF, Yang N, Gall K, McDowell DL, Fan J, Gullett PM. High cycle fatigue of die cast AZ91E-T4 magnesium alloy. *Acta Mater* 2004; 52: 1327-36

- [17] Khan AS, Pandey A, Gnäupel-Herold T, Mishra RK. Mechanical response and texture evolution of AZ31 alloy at large strains for different strain rates and temperatures. *Int J Plast* 2011; 27(5): 688-706
- [18] Morita S, Ohno N, Tamai F, Kawakami Y. Fatigue properties of rolled AZ31B magnesium alloy plate. *Trans Nonferrous Met Soc China* 2010; 20: 523-6
- [19] Ishihara S, Nan Z, Goshima T. Effect of microstructure on fatigue behavior of AZ31 magnesium alloy. *Mater Sci Eng A* 2007; 468-470: 214-22.
- [20] Sajuri ZB, Miyashita Y, Hosokai Y, Mutoh Y. Effects of Mn content and texture on fatigue properties of as-cast and extruded AZ61 magnesium alloys. *Inter J Mech Sci* 2006; 48: 198-209.
- [21] Kaneko J, Sugamata M, Numa M, Nishikawa Y, Takada H. Effect of texture on the mechanical properties and formability of magnesium wrought materials. *J Jpn Inst Met* 2000; 64(2): 141-7.
- [22] Watanabe H, Takara A, Somekawa H, Mukai T, Higashi K. Effect of texture on tensile properties at elevated temperatures in an AZ31 magnesium alloy. *Scripta Mater* 2005; 52: 449-54.
- [23] Somekawa H, Mukai T. Effect of texture on fracture toughness in extruded AZ31 magnesium alloy. *Scripta Mater* 2005; 53: 541-5.
- [24] Begum S, Chen DL, Xu S, Luo AA. Effect of strain ratio and strain rate on low cycle fatigue behavior of AZ31 wrought magnesium alloy. *Mat Sci Eng A* 2009; 517(1-2): 334-43
- [25] Begum S, Chen DL, Xu S, Luo AA. Low cycle fatigue properties of an extruded AZ31 magnesium alloy. *Int J Fatigue* 2009; 31(4): 726-35
- [26] Xiong Y, Yu Q, Jiang Y. Multiaxial fatigue of extruded AZ31B magnesium alloy. *Mat Sci Eng A* 2012; 546: 119-28
- [27] Wu L, Jain A, Brown DW, Stoica GM, Agnew SR, Clausen B, Fielden DE, Liaw PK. Twinning-detwinning behavior during the strain-controlled low-cycle fatigue testing of a wrought magnesium alloy, ZK60A. *Acta Mater* 2008; 56(4): 141-7
- [28] Wu W, Lee SY, Paradowska AM, Gao Y, Liaw PK. Twinning-detwinning behavior during fatigue-crack propagation in a wrought magnesium alloy AZ31B. *Mater Sci Eng A* 2012; 556: 278-86
- [29] Morita S, Ohno N, Tamai F, Kawakami Y. Fatigue Crack Propagation Behavior of Textured Polycrystalline Magnesium Alloys. *Mat Trans* 2010; 51 (9): 1543-1546
- [30] Morita S, Fujiwara S, Hori T, Hattori N, Somekawa H, Mayama T. Microstructure dependence of fatigue crack propagation behavior in wrought magnesium alloy. *Frattura ed Integrità Strutturale* 2016; 35: 82-87
- [31] Pook LP, Greenan AF. Fatigue crack-growth characteristics of two magnesium alloys. *Eng Fract Mech* 1973; 5(4): 935-46.
- [32] Zeng RC, Xu YB, Ke W, Han EH. Fatigue crack propagation behavior of an as-extruded magnesium alloy AZ80. *Mater Sci Eng A* 2009; 509: 1-7
- [33] Zeng RC, Ke W, Han EH. Influence of load frequency and ageing heat treatment on fatigue crack propagation rate of as-extruded AZ61 alloy. *Int J Fatigue* 2009; 31: 463-7.

- [34] Zheng, S., Yu, Q., Gao, Z., Jiang, Y. Loading history effect on fatigue crack growth of extruded AZ31B magnesium alloy. *Eng Fract Mech* 2013; 114: 42-54
- [35] Zhao T, Zhang J, Jiang Y. A study of fatigue crack growth of 7075-T651 aluminum alloy. *Int J Fatigue* 2008; 30(7): 1169-1180
- [36] Wang X, Gao Z, Zhao T, Jiang Y. An experimental study of the crack growth behavior of 16MnR pressure vessel steel. *J Pres Ves Tech* 2009; 131: 1-9
- [37] Kalnaus S, Fan F, Jiang Y, Vasudevan AK. An experimental investigation of fatigue crack growth of stainless steel 304L. *Int J Fatigue* 2009; 31: 840-849
- [38] Tada H, Paris PC, Irwin GR. The stress analysis of cracks handbook. 3rd ed. ASME International; 2000
- [39] Uematsu Y, Kakiuchi T, Nakajima M, Nakamura Y, Miyazaki S, Makino H. Fatigue crack propagation of AZ61 magnesium alloy under controlled humidity and visualization of hydrogen diffusion along the crack wake. *Int J Fatigue* 2014; 59: 234-243
- [40] Walker K, The effect of stress ratio during crack propagation and fatigue for 2024-T3 and 7075-T6 aluminum. *Effects of Environmental and Complex Loading History on Fatigue Life*, ASTM STP 462, Philadelphia, USA, 1970; 1-14
- [41] Zheng LL, Gao YF, Lee SY, Barabash RI, Lee JH, Liaw PK. Intergranular evolution near fatigue crack tips in polycrystalline metals. *J Mech Phys Solids* 2011; 59: 2307-22
- [42] Dong S, Yu Q, Jiang Y, Dong J, Wang F, Ding W. Electron backscatter diffraction observations of twinning-detwinning evolution in a magnesium alloy subjected to large strain amplitude cyclic loading. *Mater & Design* 2015; 65: 762-765
- [43] Ando S, Iwamoto N, Hori T, Tonda H. Fatigue crack propagation in magnesium crystals. *J Jpn Inst Met* 2001; 65(3): 187-190
- [44] Ando S, Saruwatari K, Hori T, Tonda H. Fatigue crack propagation behavior in magnesium single crystals. *J Jpn Inst Met* 2003; 67(5): 247-251
- [45] Ando S, Ikejiri Y, Iida N, Tsushida M, Tonda H. Orientation dependence of fatigue crack propagation in magnesium single crystals. *J Jpn Inst Met* 2006; 70(8): 634-637