Microstructure-based multistage fatigue modeling of a cast AE44 magnesium alloy

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Abstract

The multistage fatigue model developed by McDowell et al. was modified to study the fatigue life of a magnesium alloy AE44 for automobile applications. The fractographic examination indicated three distinct stages of fatigue damage in the high cycle fatigue loading regime: crack incubation, microstructurally small crack growth, and long crack growth. Cracks incubated almost exclusively at the cast pores that were near the free surface, located near sharp geometry changes of the test specimen, or at extremely large pores inside the specimen. Microstructurally small cracks grew in the eutectic region along the weak boundaries of the grains and dendrites or at very closely packed microstructural discontinuities. Long cracks were observed to grow in a transgranular fashion. Specimens fabricated from as-cast bars and extracted from a cast engine cradle were tested at room temperature and an elevated temperature typically required for automotive powertrain applications. A large variation of fatigue life in the high cycle fatigue region was observed in specimens from both conditions due to the sensitivity from microstructural discontinuities. The microstructure-based multistage fatigue model was generalized for the AE44 magnesium alloy to capture the network of porosity and temperature dependence. The modified multistage fatigue model was also used to estimate the upper and lower bounds of the strain-life curves based on the extreme microstructural discontinuities.

Keywords: Cast magnesium alloy; Fatigue modeling; Fatigue experiments; Fractography; Temperature effects

1. Introduction

Magnesium alloys have received increased interest for automotive applications because of their attractive low density, the possible benefit of weight reduction, and consequent reduction in fuel consumption. The rather high specific strength, the good castability, the improved corrosion resistance of high purity alloys, and the possibility of recycling are additional advantages of these materials [1–6]. Cast magnesium alloys are used in several automotive applications such as covers, door structures, and heavier load-bearing components like wheels, housings and frames. Vehicle components are subjected to variable and repeated mechanical straining, where the number of load cycles may be on the order of $10^9$ cycles in car wheels. When magnesium alloys are considered for powertrain components, such as crankcases or gearbox casings, fatigue properties at elevated temperatures are of great interest. The effects of the processing methods on the microstructures, mechanical properties, fracture and fatigue behavior of cast Mg alloys have been reported in the open literature [7–14]. Earlier fatigue investigations have shown that porosity plays a key role in crack initiation locations in cast Mg [15–17]. Fatigue studies of AM50 Mg alloy samples obtained from different locations of prototype control arms and wheels showed that the fatigue performance was significantly affected by shrinkage porosity and gas pores, depending on the location on the components where samples were extracted [18]. Design related fatigue experiments and

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estimations for typical applications were conducted on smooth and notched specimens [19] and under constant and variable loading cases [20].

In this paper, the fatigue behavior of an AE44 Mg alloy is explored using test samples extracted from as-cast bars and cast automotive components. The strain-life behavior of the AE44 magnesium alloy was studied for two types of specimens at room temperature and at 121°C/176°C, a temperature range typical of automotive powertrain applications. The as-cast bars have a more consistent porosity level from specimen to specimen but can have large casting pores. The average dendrite cell size (DCS) for as-cast bars is small and relatively uniform from specimen to specimen. The engine cradle specimens have a wider variety of defect discontinuities from the thin section to the thick section of the cradle and can have very large pores. The engine cradle cell size (DCS) for as-cast bars is small and relatively uniform from specimen to specimen. The engine cradle specimens have a wider variety of defect discontinuities from the thin section to the thick section of the cradle and can have very large pores.

The microstructure-based multistage fatigue (MSF) model of McDowell et al. [21] was generalized for the AE44 Mg alloy. This model recognized the multiple inclusion severity scales for crack formation and addressed the role of constrained microplasticity around debonded particles or casting pores in forming and growing microstructurally small fatigue cracks. The “base pore size” for incubation for AE44 Mg alloy was estimated as the average pore size of one of a million pores in the large tail end of the pore distribution. Similar to the MSF model, the incubation was assumed to consume over 90% of the total cycles to failure in the high cycle fatigue (HCF) regime and approximately 30% of the total cycles to failure at the low cycle fatigue (LCF) regime. A few numerical constants in the MSF model were estimated based on their physical significance in the incubation model. A new factor is introduced to capture the effect of local porosity on acceleration of the microstructurally small crack (MSC) growth. The modified microstructure-based multistage fatigue model was employed to capture the upper and lower bounds of the fatigue life based on the limits of the microstructural discontinuities.

2. Material and experiments

2.1. Materials

The Mg AE44 alloy was fabricated using a high-pressure die cast technique. Test specimens were fabricated from as-cast bars and from cast engine cradles. The chemical composition was measured by means of an Electron Probe MicroAnalysis (EPMA) and the results are reported in Table 1. Also shown is the nominal chemical composition of the alloy provided by General Motors (GM) Corporation. The distribution of the elements along the microstructural features of the alloy was examined using X-ray mapping of the EPMA. The 4% Al was primarily concentrated in the eutectic region as a compound with rare earth elements of the form Al,RE, at the dendrite boundaries inclosing the α-phase magnesium matrix. Mn and Si were in the formations of intermetallic particles scattered in the β-phase magnesium matrix.

Table 1
Composition of cast AE44 magnesium alloy provided by GM and measured by means of EPMA

<table>
<thead>
<tr>
<th>Resource</th>
<th>Mg</th>
<th>Al</th>
<th>Mn</th>
<th>Zn</th>
<th>Si</th>
<th>Fe</th>
<th>RE</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>GM</td>
<td>Bal.</td>
<td>3.5–4.5</td>
<td>0.15–0.5</td>
<td>0.2max</td>
<td>0.1max</td>
<td>0.005max</td>
<td>3.5–4.5</td>
<td>None</td>
</tr>
<tr>
<td>EPMA</td>
<td>87.88</td>
<td>3.44</td>
<td>0.18</td>
<td>0.015</td>
<td>0.005</td>
<td>0.0038</td>
<td>1.9 (Ce)</td>
<td>0.88</td>
</tr>
</tbody>
</table>

All quantities are in weight percent.
2.2. Specimen

Smooth cylindrical dog-bone shaped fatigue specimens were employed based on ASTM standards (E-747-99). Standard sub-size specimens were fabricated with a gauge length of 25 mm and a uniformed gage cross-section diameter of 4.7 mm. The cross-section at the grip was 9.4 mm in diameter, and the total length of the specimens was 90 mm. The specimen cross-section was evaluated using an optical microscope and a backscatter field emission gun (FEG) scanning electron microscope (SEM). Fig. 1 shows the optical micrographs of polished sample cross-sections that illustrate the size and distribution of gas pores, which demonstrates that the gas pore size was larger and the shrinkage porosity was higher in the sample obtained from the engine cradle as compared to the as-cast bars. A high magnification optical micrograph of the same polished cross-sections revealed the morphology of the dendrite cells and boundaries, as shown in Fig. 2. A distinct dendrite size variation from the as-cast bar to thin sections of the engine cradle and thicker sections of the engine cradle was evident. The dendrite cell size (DCS) was measured employing the linear intercept method (ASTM E112) using AxioVision Grains software. The DCS of the as-cast bars varied from 2 to 10 μm over the whole cross-section with an average of 5 μm. The DCS of samples machined from the engine cradle increased from the thinner to thicker cross-sections varied from 9–15 to 12–17 μm with an average of 15 μm. The primary DCS in a thin section of the cast engine cradle was about 13% less than the DCS in a thick section.

2.3. Experimental procedures

The experiments were conducted using an MTS servo-controlled electro-hydraulic system. Monotonic and cyclic hardening experiments were conducted using the same specimens and test setup as the fatigue tests. The tests were conducted under strain-controlled, constant strain amplitude conditions, with the strain measured using a 0.5 in. axial fatigue rated extensometer attached within the gage length. Cyclic loading was applied at frequencies of 0.5 Hz for the first 43,000 cycles and 20 Hz, thereafter. A 50% drop in the peak cyclic load was used to determine the final failure of the specimens according to ASTM standards. A temperature chamber was employed for a constant elevated temperature during the fatigue test at 121 °C. The strain amplitudes ranged 0.05% strain to above the yield strength 0.6%. The remotely applied strain ratio of $R_e = e_{\text{min}}/e_{\text{max}} = -1$ was employed in all experiments. At least three replicated tests were conducted for each loading condition.

3. Experimental results

3.1. Static and fatigue

Uniaxial tension tests were conducted at room temperature and 121 °C with a loading rate of 0.005 min⁻¹. The stress–strain curves are shown in Fig. 3. The elastic modulus, yield strength, ultimate strength and strain-to-failure of the as-cast AE44 bars were higher than those extracted from the engine cradle. The static and cyclic properties are shown in Table 2. The Young’s modulus and the yield
strength of the as-cast bars were approximately 14% and 19% higher than the corresponding properties of the cradle, due to the cradle's large pore sizes and high porosity. The uniaxial strain-life results are shown in Fig. 4. A large variation of fatigue lives in the HCF regime was observed on specimens from both conditions, due to the sensitivity to microstructure. The large scatter in the HCF regime could be induced by the sub-sized ASTM specimens in which the above stated detrimental pore was almost one tenth of the specimen diameter. The fatigue lives at 121°C are longer, in general, than those at the room temperature under the constant strain amplitude loading condition with the strain-controlled experiments, which maybe is induced by the increasing ductility at the elevated temperature. In addition, the specimens experienced large cyclic plastic strains when the strain amplitudes were greater than 0.28%. As expected, the cyclic plastic strain amplitudes in the cradle specimens were higher than those in the as-cast bars for the same applied strain amplitudes. However, the cradle specimens exhibited higher fatigue lives than those from the as-cast bars at the LCF region.

Table 2
Mechanical properties of AE44 magnesium alloy in (1) as-cast bar and (2) extracted from engine cradle at room and 121 °C

<table>
<thead>
<tr>
<th>Properties</th>
<th>AE44-Bar</th>
<th>AE44-Cradle</th>
<th>AE44-Bar at 121 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Elastic modulus (MPa)</td>
<td>44,740</td>
<td>38,620</td>
<td>40,830</td>
</tr>
<tr>
<td>Yield strength (MPa)</td>
<td>135</td>
<td>109</td>
<td>112</td>
</tr>
<tr>
<td>Ultimate strength (MPa)</td>
<td>244</td>
<td>182</td>
<td>160</td>
</tr>
<tr>
<td>Cyclic strength</td>
<td>624</td>
<td>685</td>
<td>635</td>
</tr>
<tr>
<td>Coef. ((K^c) MPa)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cyclic strain hardening exp. ((n^c))</td>
<td>0.32</td>
<td>0.30</td>
<td>0.351</td>
</tr>
<tr>
<td>Cyclic yield strength (MPa)</td>
<td>89</td>
<td>106</td>
<td>75</td>
</tr>
<tr>
<td>Fatigue ductility exp. ((c))</td>
<td>(-0.62)</td>
<td>(-0.52)</td>
<td>(-0.505)</td>
</tr>
<tr>
<td>Fatigue ductility coef.</td>
<td>0.26</td>
<td>0.10</td>
<td>0.1</td>
</tr>
<tr>
<td>Cycles to failure, (N_f)</td>
<td>1e+2</td>
<td>1e+3</td>
<td>1e+4</td>
</tr>
</tbody>
</table>

Fig. 2. Optical images from polished cross-sections present an overview of morphologies of dendrite cells and boundaries in the AE44 magnesium alloy: (a) as-cast bar and (b) from an engine cradle.

Fig. 3. Stress–strain curves of the AE 44 magnesium alloy: (1) as-cast bar, (2) extracted from engine cradle tested at room temperature (RT) and 121 °C (HT).

Fig. 4. Uniaxial strain-life of AE44 Mg alloy specimens machined from as-cast bars and from an engine cradle tested at room temperature (RT) and 121 °C (HT) under strain-controlled, constant amplitude with completely reversed strain amplitude experiments.
3.2. Fractographic analyses

Typical specimens loaded in the strain amplitude of 0.075% (HCF) and 0.3% (LCF) were selected for fractographic analyses. At $\Delta e/2 = 0.075\%$, some specimens did not fail at $10^7$ cycles. A sample of medium life, S1 ($N_f = 1.3 \times 10^6$, at 121 °C), and samples with the shortest lives, S2 ($N_f = 1.78 \times 10^5$, at 121 °C), and S3 ($N_f = 1.83 \times 10^5$, at 25 °C) were examined under the SEM, which are shown in Fig. 5. An arrow indicates the location of the fatigue crack formation site. Most cracks formed at casting pores at or near the specimen surfaces or at a very large pore within the specimen as shown in Fig. 6. A strong interaction of two pores that were about 25 μm in diameter near the free surface formed a fatigue crack that produced the fatigue failure.

A few relatively larger pores were scattered inside this specimen but no damage was formed. Even though S2 and S3 had similar low fatigue lives, the life incubation sites were quite different: the damage in S2 formed at the interaction of two pores that were 65 μm in diameter with one right at the edge while the damage in S3 formed at a large pore of 500 μm in diameter in the middle of the specimen. Therefore, the pore sizes of 25 μm and 500 μm were used to demonstrate the upper and lower bounds of the fatigue lives.

Fig. 5. SEM micrographs on the overall fracture surfaces of samples under 0.075% strain amplitude: (a) S1: $N_f = 1.3 \times 10^6$ at 121 °C; (b) S2: $N_f = 1.78 \times 10^5$ at 121 °C; (c) S3: $N_f = 1.83 \times 10^5$ at room temperature. The locations of the incubation sites are indicated by arrows.
The backscatter electron images on the sample S1 on the earlier crack growth region indicated Al\textsubscript{x}RE\textsubscript{y}-rich dendrite boundaries with the heavier Al\textsubscript{x}RE\textsubscript{y} compound marked as white in Fig. 7. Fig. 7a shows that the fracture surface comprised regions of dendrite boundaries, which manifested the crack propagating and meandering along the dendrite boundaries with the combination of interdendritic and intradendritic crack growth. The large striations on the dendrite boundaries indicate the fast segregation along the boundaries MSC and earlier LC growth.

Further into the specimens, the fracture surface became rough and intragranular and intergranular crack growth was observed as shown in Fig. 8b. The MSC grew in the eutectic region along the weak grain and dendrite boundaries under the influences of very closely packed microstructural discontinuities. The eutectic region comprises weak bonding regions with accompanied cast pore and shrinkage porosity. In this region, the load bearing capability drops due to the large porosity, and simultaneous crack propagation is accelerated by the weak bonding between dendrites, which also enhances the scatter in fatigue life. Long cracks were observed to grow in a transgranular fashion and the fracture surfaces are relatively flat scattered with dimples formed from the pores.

Fig. 6. SEM images of fatigue crack formation sites for samples: (a) S1: interaction of two casting pores of approximately 25 μm diameter near the edge; (b) S2: interaction of two cast pores of approximately 60 μm diameter at the edge; (c) S3: a large casting pore of approximately 0.5 mm in a linear dimension at the center of the specimens.
4. Multistage fatigue model

A microstructure-based multistage fatigue (MSF) model that incorporates different microstructural discontinuities (pores, inclusions, etc.) on physical damage progression was implemented to model the fatigue life of the AE44 Mg alloy. This model partitions the fatigue life into three stages based on the fatigue damage formation and propagation mechanisms: crack incubation, microstructurally small crack (MSC) and physically small crack (PSC) growth, and long crack (LC) growth. This model was initially developed for a cast A356-T6 [21] Al alloy and modified and generalized for a variety of aluminum alloys [22, 23]. As stated above, three types of defects were listed in descending order of severity with regard to detriment of fatigue life: (1) gas pores greater than 25 μm in diameter; (2) shrinkage porosity clusters with special dendrite features greater than 300 x 300 μm² in size that demonstrated locally high porosity, and (3) extreme large pores of 500 μm or greater in diameter. In most cases, the shrinkage porosity was located in the eutectic region, weakening the load bearing capability and thereby enhancing the propagation rate of interdendritic cracks, leading to a greater reduction in fatigue lives.

A brief recast of the multistage fatigue model is described in the following accompanied by the detailed explanation of model parameters for the AE44 Mg alloy. The total fatigue life is decomposed into the cumulative number of cycles spent in several consecutive stages as follows:

\[ N_{\text{Total}} = N_{\text{Inc}} + N_{\text{MSC/PSC}} + N_{\text{LC}}, \] (1)
where $N_{\text{inc}}$ is the number of cycles to incubate a crack at a micronotch that includes the nucleation of crack-like damage and early crack propagation through the region of the micronotch root influence. $N_{\text{MSC}}$ is the number of cycles required for propagation of a microstructurally small crack with the crack length, $a < a < k DCS$, with the DCS defined as the dendrite cell size, and $k$ as the non-dimensional factor that is representative of a saturation limit for the encountering of a 3-D crack front with sets of microstructural discontinuities. $N_{\text{PSC}}$ is the number of cycles required for propagation of a physically small crack (1–2) DCS < $a < O (10 \text{DCS})$, during the transition from MSC status to that of a dominant LC. Depending on the microstructural inclusion morphology and texture of the matrix, the PSC regime may conservatively extend to 300–800 μm. Also, the scale of the crack tip plastic zone is typically small for the PSC regime that it samples individual microstructural elements; hence, there is an insufficient volume of material to achieve the length scale-independent afforded by homogeneous $\Delta K$ solutions in long crack fracture mechanics. In Eq. (1), the MSC and PSC regimes were combined into one mathematical form. $N_{\text{LC}}$ is number of cycles required for LC propagation for crack length $a > (10–20) \text{ DCS}$, depending on the amplitude of loading and the corresponding extent of microplasticity ahead of the crack tip. This stage of crack extension is commonly characterized using standard fatigue crack growth experiments, $da/dN$ versus $\Delta K$.

4.1. Incubation

The incubation of a fatigue crack at a local discontinuity or a stress raiser was induced by the notch root microplasticity. Gas pores are life-limiting inclusions for AE44 Mg alloy, being relatively large in size and located near the free surface or near a sharp geometric change in a specimen. Using digital image analysis, the probability distribution of pore size was estimated in a one-millimeter by one-millimeter square area. The estimation of the life-limiting pore size corresponds to only one of a million pores in the large tail end of the distribution that will contribute to the incubation of the fatigue crack for smooth specimens considered with a uniform stress state.

The correlation of the local plastic zone size and the macroscopic applied strain amplitude can be obtained using micromechanical finite element simulations. The correlation was obtained for the A356 Al alloy by Gall et al. [24], i.e.,

$$\frac{l}{D} = \frac{\langle \varepsilon_{\text{a}} - \varepsilon_{\text{th}} \rangle}{\varepsilon_{\text{per}} - \varepsilon_{\text{th}}}, \quad \frac{l}{D} \leq \eta_{\text{lim}},$$

$$(2a)$$

$$\frac{l}{D} = 1 - (1 - \eta_{\text{lim}}) \left( \frac{\varepsilon_{\text{per}}}{\varepsilon_{\text{a}}} \right)^{r}, \quad \eta_{\text{lim}} \leq \frac{l}{D} \leq 1,$$

$$(2b)$$

where function $\langle f \rangle = \int f \, df$ if $f \geq 0$; $\langle f \rangle = 0$ otherwise. Also, “$r$” is a shape constant for the transition to limit plasticity and $\eta_{\text{lim}}$ is the linear factor [21]. Parameters $\varepsilon_{\text{th}}$ and $\varepsilon_{\text{per}}$ are material constants designating the fatigue damage threshold and the strain percolation limits, respectively, for microplasticity. The two parameters can be estimated using macroscopic material properties. $\varepsilon_{\text{a}}$ is the macroscopic remote cyclic strain amplitude. $l$ is the plastic zone size in front of the inclusion projected to the direction perpendicular to the loading, and $D$ is the linear dimension of the inclusion particle that incubated the crack that caused the fatigue failure. The ratio $l/D$ was used to estimate of the notch root plasticity at the influence of the particle, with the limit factor $\eta_{\text{lim}}$ indicating the transition from constrained microplasticity to unconstrained microplasticity at the micronotch. $r = 0.2$, $\eta_{\text{lim}} = 0.3$ [21] were found also suitable for the AE44 Mg alloy.

To study the damage incubation life as a function of local plastic deformation, a modified Coffin–Manson law was implemented based on the non-local maximum plastic shear strain, i.e.,

$$C_{\text{inc}} N_{\text{inc}}^{2} = \beta = \frac{\Delta \gamma_{p}}{2},$$

$$(3)$$

where $\beta$ is the non-local maximum plastic shear strain amplitude around the inclusion calculated using an average of maximum plastic shear strain over an area approximately one percent of the inclusion area; $C_{\text{inc}}$ and $\alpha$ are the linear and exponential coefficients in the modified Coffin–Manson law at the micronotch. Exponent $\alpha$ is similar to that in the macroscopic Coffin–Manson Law for strain life; coefficient $C_{\text{inc}}$ is highly dependent on the remote load ratio in the CHF region. It is assumed that $C_{\text{inc}} = c (1 - R)$ [21], where $c$ is material dependent constant that equals 0.24 for AE44 Mg alloy. The coefficient is dependent on the local plastic zone size in the transition to the unconstrained microplasticity, thus it becomes a function of the remote strain amplitude. The non-local maximum shear strain at the micronotch can be estimated using micromechanics analysis of a representative volume cell with a non-local averaging procedure [24]

$$\beta = \frac{\Delta \gamma_{p \text{max}}}{2} = Y [ (\varepsilon_{\text{a}} - \varepsilon_{\text{th}}) ]^{q}, \quad \frac{l}{D} \leq \eta_{\text{lim}},$$

$$(4a)$$

$$\beta = \frac{\Delta \gamma_{p \text{max}}}{2} = Y \left( 1 + \xi \frac{l}{D} \right) [ (\varepsilon_{\text{a}} - \varepsilon_{\text{th}}) ]^{q}, \quad \eta_{\text{lim}} \leq \frac{l}{D} \leq 1,$$

$$(4b)$$

where $Y$, $q$, $\eta_{\text{lim}}$ are material and microstructure-related parameters obtained using micromechanical simulation at the inclusion. Parameter $Y$ depends on the remote load ratio; based on finite element analysis for A356 Al alloy, it was assumed $Y = [y_{1} + (1 + R) y_{2} (2D/D_{0})^{r} ]$, with $y_{1}$ and $y_{2}$ as constants. $D_{0}$ is the diameter of the pore in the specimens that caused fatigue failure and the exponential $d'$ depicts the effect of pore size to local plastic strain, which is taken as $0.1 < d' < 0.6$.

The strain threshold can be estimated using the fatigue endurance limit that is measured using the standard stress-life fatigue experiment in the HCF regime. The ultimate strength of the metallic alloys can be used to estimate the strain threshold $\varepsilon_{\text{th}} = 0.285 S_{u}/D$. The fatigue strength at
$10^7$ under completely reversed strains can also be used to estimate the fatigue strength amplitude at $10^7$ cycles, a procedure recommended by Maenning [26] gave the probability for failure, i.e.,

$$P_F(e_a) = \left[ \frac{3i - 1}{3n + 1} \right]^{\frac{i}{3n + 1}},$$  

(5)

where $i$ is the number of specimens failed below $10^7$ cycles, and $n$ is the number of specimens tested at the particular strain amplitude. Stress controlled tests are usually used with at least 25 specimens to estimate the fatigue strength. The percolation limit for microplasticity denotes that the crack growth rate is linearly proportional to the load bearing capacity of the matrix, $g(\phi) = (1 - \phi)^{-\zeta}$ and $0 < \zeta < 1$. The parameter $U$ is defined as the load ratio parameter in the form of $U = \frac{1}{\sqrt{P}}$. More detailed dependence on microstructure can be built into the MSC growth relation as required. The macroscopic maximum plastic shear strain amplitude is calculated using $\varepsilon_{\text{Y}}/\varepsilon_c = \frac{1}{2} \left( \frac{\Delta \varphi}{\Delta \varphi} \right)^2$ for the uniaxial loading case.

Finally, the LC regime is represented by linear elastic fracture mechanics (LEFM) growth,

$$\left( \frac{da}{dN} \right)_{\text{LC}} = A([\Delta K_{\text{eff}}])^{m} - ([\Delta K_{\text{eff,th}}])^{m},$$  

(9)

where the intrinsic threshold and the effective stress intensity factor range are used, and $A$ and $m$ are the crack growth material constants in the Paris Law [27]. The effective stress intensity factor range is defined by $\Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{op}}$ if $K_{\text{min}} < K_{\text{op}}$, $\Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{min}}$ if $K_{\text{min}} \geq K_{\text{op}}$. The opening stress intensity factor level can be calculated using finite element analysis based on opening stresses at the crack tip. The stress intensity factor is given by

$$K = f \left( \frac{da}{dN} \right) \sqrt{\pi a},$$  

(10)

where $a$ is the radius of the semi-circular crack, and $d$ is the diameter of the cylindrical specimen. The geometry function for a semi-circular crack at the surface of a cylindrical specimen for mode I fracture can be written as

$$f(x) = 0.67 - 1.24(x) + 28.0(x)^2 - 162.4(x)^3 + 472.2(x)^4 - 629.6(x)^5 + 326.1(x)^6.$$  

(11)

The transition between the MSC and LC growth was governed by selecting the maximum of either of the two rates, i.e.,

$$\frac{da}{dN} = \max \left[ \left( \frac{da}{dN} \right)_{\text{MSC}}, \left( \frac{da}{dN} \right)_{\text{LC}} \right].$$  

(12)

5. Multistage fatigue model correlation and fatigue life estimation

The threshold strain amplitudes for incubating fatigue damage were estimated using the ultimate strength of the alloys that are listed in Table 2, i.e., $\varepsilon_{\text{th}} = 0.08\%, 0.07\%$, respectively, for samples from the as-cast bar and the crack at room temperature, and $\varepsilon_{\text{th}} = 0.06\%$ for the as-cast bar at 121 °C. The threshold strain amplitude calculated from fatigue strength at $10^7$ cycles with 50% probability as estimated using Eq. (5) was estimated to be between
0.05% and 0.075% for the AE44 Mg alloys, based on a limited number of experiments. These values are much less than those obtained for other similar cast Mg alloys [29,30], which could be caused by the large casting pores in the sub-sized ASTM specimens. The percolation limit for microplasticity at the notch was estimated directly using the cyclic yield strengths that are listed in Table 2. The cycles spent incubating a crack and propagating it through the micronotch root plasticity affected zone was estimated using Eq. (6). It was assumed that the non-local plastic shear strain amplitudes and local plastic zone size holds the same form as a function of applied strain amplitude and the $R$-ratio obtained for A356 [21]. The material constants for incubation were correlated using a mean square root regression scheme. In the MSC growth region, the crack tip driving force was estimated using the crack tip displacement that included two parts of contribution: the first term in Eq. (8) is primarily dominant in the HCF regime and the second term is dominant in the LCF regime. The primary contributions and the interaction of the two terms were treated as weighted constraint factors in the model correlation to obtain the constants $C_I$ and $C_{II}$.

Figs. 9 and 10 show the multistage fatigue model correlations for AE44 magnesium alloy in the as-cast and cradle specimens at room and elevated temperatures, respectively. Assuming a large pore size of 500 µm as the site of damage incubation and high porosity near the pore, the lower bound of fatigue life of AE 44 in the as-cast condition was estimated. Assuming an extremely small pore size of 10 µm as the site of damage incubation and no abnormal porosity near the pore, the upper bound for fatigue lives of AE44 as-cast bar specimens were estimated as shown in Fig. 11. The capability of estimating correct upper and low bounds demonstrates the robustness of the multistage fatigue model, which includes the key microstructure features that affect the fatigue life.

6. Conclusions

Samples of high-pressure die cast AE44 magnesium alloys extracted from as-cast bars and from an engine cradle were tested until failure under strain control at room and elevated temperatures. The initial microstructure and fatigue fractographs were examined using optical microscopy and SEM. A multistage fatigue model originally proposed by McDowell et al. [21], was modified for greater porosity interaction and temperature dependence. The model captured the following microstructural phenomena in fatigue crack growth:

(1) Fatigue damage formed at large cast pores near the free surfaces with an average size greater than 25 µm; the most detrimental microstructural discontinuity to fatigue life were the largest pores greater than 400 µm and the interaction between large pores.
(2) The multistage fatigue model was able to explain the extreme limiting conditions for the strain-life of the AE44 Mg alloy. The upper and lower bounds for fatigue life were estimated using observed extreme microstructure features that coincided with the experimental results.

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References