Research Article

Fatigue Crack Propagation Behavior of the 2195-T8 Aluminum-Lithium Alloy with a Precorroded Hole

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This work investigated the fatigue crack propagation behavior of the 2195-T8 Al-Li alloy with precorroded holes. The microstructure, crack growth path, and the fracture surfaces of the samples with the precorroded holes have been examined with scanning electron microscopy (SEM), electron backscatter diffraction (EBSD), and computed tomography (CT). The characteristics of the crack growth have been analyzed via fracture mechanics by employing the finite element method (FEM). The results showed that the bulk of the fatigue cracks had been initiated from the corrosion pits and corroded discontinuously and precipitated further to become intergranular. Owing to the superposed stress concentration caused by the corrosion pits, the porous region at the bottom of the precorroded hole had affected the crack propagation performance. Crack growth rates were different in multiple positions owing to the difference of \( K_I \) at crack fronts.

1. Introduction

The combination of the properties of low density and high strength of Al-Li alloys provides potential applications in aerospace industry, thus resulting in the weight reduction and performance improvement of launch vehicles and missiles [1–3]. As metallic materials for aviation industry, their fatigue performance and resistance have tremendous influence on the component life [4–6].

Research on the fatigue fracture of aluminum alloys has been performed from multiple perspectives. The initiation of microcracks [7–9], influence factors of growth [10], crack propagation path [11], evolution of crack surface, and environmental conditions of fatigue [12, 13] are currently pivotal study domains. Mingzhe et al. [14] and Srivatsan et al. [15] have reported that most of the cracks initiated from the interface between the coarse particles and the matrix or inclusions. Those fractured particles have been considered as prior fatigue cracks owing to the immense stress concentration, and their arrangement modes could change the fatigue crack propagation rates [16, 17]. It has been reported that the crack propagation direction could be influenced by the misorientations among the grains and grain boundaries (GBs). Zhai et al. [18, 19] have proposed a model, including the twist and tilt angles, to describe the barrier of GBs for the crack growth in the crystallographic planes. Jian et al. [20] have revealed the effect of the interfacial angle between two adjacent grains on the intergranular propagation and crack deflection, controlled by the grain orientations, and it has been confirmed by Mingzhe et al. [14].

Owing to the corrosion-enhanced cracking behavior, the fatigue characteristics affected by corrosion have been extensively studied. Microobservation has shown that the corrosion pits caused the multisite crack sources and reduced the time of fatigue crack nucleation which decreased the fatigue life [21, 22]. Song et al. [23] have demonstrated that the crack deflection and coalescence were related to the corrosion pits. Kim et al. [24] have found cracks being initiated from the pits that could be regarded as semielliptical surface micronotches. Based on observation, Walde and Hillbery [25] and Xiang et al. [26] have developed several equivalent initial flaw models to transform defects into simplified cracks further to better predict fatigue life. Co and Burns et al. [27] have employed a crack surface marking method.
to calculate the microcrack growth and concluded that the crack growth rate was independent of the corrosion damage morphology. Furthermore, Turnbull et al. [28], Amiri et al. [29], and Hu et al. [30] have used various methods, including experimental observation, continuum damage mechanics, and elastic-plastic damage evolution model, to study the pit-to-crack process.

In practice, the defects always appear on the surface of the alloys due to various factors. They can cause further stress concentration to initiate cracks and significantly lower the service life of the equipment. However, the spatial and systematic characterization of the fatigue crack in 2195 Al-Li alloys, with precorrroded defects, has not been studied adequately. As a type of propellant tank material in N$_2$O$_4$ environment, the 2195 Al-Li alloys would experience corrosion to a certain extent. Under the role of water vapor, N$_2$O$_4$ would transform to HNO$_3$, adding to the toxicity of N$_2$O$_4$, and therefore, precorrroded defects, has not been studied adequately. As a type of propellant tank material in N$_2$O$_4$ environment, the 2195 Al-Li alloys would experience corrosion to a certain extent. Under the role of water vapor, N$_2$O$_4$ would transform to HNO$_3$, adding to the toxicity of N$_2$O$_4$, and therefore, precorrroded environment is set to be HNO$_3$ [31, 32].

This work investigated the fatigue crack propagation behavior in 2195-T8 Al-Li alloys with a precorroded hole. Microscopic observation and 3D reconstructed model have been used to reveal characteristics of fatigue crack growth, which were preceded by precorrosion and followed by fatigue tests. Fatigue crack growth simulation in an equivalent model by LEFM approach has demonstrated the difference of the crack growth rates and the evolution of crack surface.

**2. Experimental Material and Methods**

2.1. Material and Fabrication. We investigate the properties and characteristics of 2195-T8 Al-Li alloy, in the form of 4.5 mm thick plate received from Southwest Aluminum Co. Ltd. The yield strength, ultimate strength, and elongation have been 567 MPa, 607 MPa, and 12%, respectively. The chemical composition of the alloy is listed in Table 1. The plate has been machined into dog-bone shape specimen with a length of 25 mm × 112 mm by electric wire cutting along the rolling direction. Those specimens were mechanically ground with the grid sizes from 400 to 2,000 by silicon carbide paper. Thereafter, due to easy positioning and grinding with the grid sizes from 400 to 2,000 by silicon carbide paper. Then, the specimen has been left standing in the drying oven to relieve stress. Subsequently, three specimens have been signed from #1 to #3 and immersed in acetone and cleaned by ultrasonic cleaning separately. Wax has been used to seal the section between the tube and specimen. According to Figures 2(a) and 2(b), a short plastic tube covered the hole totally to limit the corrosion area. Wax has been used to seal the section between the tube and specimen. Subsequently, nitric acid solution was dripped into the tube to corrode the hole. Thereafter, the tube mouth was sealed by wax. Such a state was prolonged for 8 h. Specimens were cleaned as described in Section 2.2 after finishing the corrosion tests. Dried specimens underwent a fatigue test at room temperature with a loading frequency of 5 Hz and load amplitude of 280 MPa at load ratio $R$ of 0.1, which was incompliance with the GBT 26077-2010 standard. The loading direction was aligned with the rolling direction. The test has been carried out by INSTRON 8801 in Aircraft Strength Research Institute of Xi’an (China).

To investigate characteristics of fatigue crack propagation, fatigue tests have been terminated manually after certain loading cycles. Computed tomography (SkysCAN2211) has been utilized to analyze the crack propagation in 3D perspective. Following the fatigue fracture, fractography was displayed using SEM with the secondary electron mode.

**3. Results and Discussion**

3.1. Microstructural Characterization. Figure 3 shows the SEM images of the morphology of the 2195-T8 Al-Li alloy in BSE mode. It can be seen that white constituent particles, which consisted of Al, Cu, Fe, and Mg elements from the EDS results were widely distributed. Clusters of particles were distributed in strips along the rolling direction on the plane, as shown in Figures 3(a) and 3(b), whereas the area between the regularly arranged particles had sparse particles. According to the magnified view in Figures 3(c) and 3(d), particles were located in grains or on GBs. Furthermore, certain minor holes, shown by arrows, which might be produced during the alloy processing and manufacturing also appeared on the surface. The existence of the constituent particles and minor defects could influence the bonding strength between the grains and fatigue strength of the alloy to a certain extent [16, 23].

Figure 4 represents the TEM micrographs and the corresponding EDX results. It showed that the needle-shaped T$_1$ (Al$_5$CuLi) distributed both in grains or GBs, whereas lath $\theta$ (Al$_5$Cu) and circular $\beta$ (Al$_3$Zr/Al$_3$Li) precipitated primarily within the grains. It has been reported

| Material  | Cu  | Li  | Si  | Fe  | Mn  | Zn  | Ti  | Zr  | Ag  | Al   |
|-----------|-----|-----|-----|-----|-----|-----|-----|-----|-----|------|
| 2195-T8   | 4.0 | 1.0 | 0.03| 0.05| 0.05| 0.01| 0.02| 0.14| 0.4 | Bal. |

To study the fatigue crack growth rate was independent of the corrosion damage morphology.
Figure 1: A specimen with a drilled hole.

Figure 2: Corrosion and fatigue tests: (a) corrosion cell, (b) detailed schematic of (a), and (c) a fatigue test.
that the coherent-lattice $\delta$ phases could improve the matrix strength and hinder recrystallization. Furthermore, there were certain Cu-rich secondary precipitates, shown in Figure 4(a) with oval lines, located on the grain boundary. These hard and brittle precipitates could affect the continuity and bonding strength between the grains, which evolved into the crack source leading to the final alloy failure easily [23, 24]. Precipitate free zones (PFZs) have not been observed for the 2195-T8 alloys when compared with other aluminum alloys, which was a reason for its high static strength.

According to the inverse pole figure (IPF) in Figure 5(a), the grains in the alloy consisted of both tiny grains and relatively large grains. Tiny grains, on the left of the IPF, were the recrystallized grains with high energy during the annealing process. Large elongated grains, on the middle of the IPF, had weak color gradient, which suggested that the deformation dislocations have been eliminated. Figure 5(b) displays local misorientations and their distribution among different grains. Local misorientations almost disappeared in large grains that were considered as subgrains, whereas misorientations, as shown in green color, still existed among those tiny ones, suggesting that annealing process needed to be improved. Moreover, majority of the local orientations were concentrated between 0° and 1°, which illustrate that the grains have been recrystallized. But recrystallization is not complete owing to the existence of the grains whose local orientations were more than 2°. Figures 5(c) and 5(d) show the point-to-point misorientation distribution along traces A and B, respectively, according to Figure 5(a). The maximum misorientation is displayed by triangular symbols, when a trace passed the grain boundary, whereas the orientations are small in large subgrains. The regions with more tiny grains had more high-angle GBs.

3.2. Surface Crack Propagation Characteristics. Figures 6(a)–6(c) display the fatigue crack propagation in the corroded hole after 22165 cycles. To ensure stress distribution of the hole under quasistatic axial loading, FEM analysis has been performed in ABAQUS. The meshing of the specimen was achieved using an 8-node brick element (C3D8). In the simulation, the Young’s modulus and Poisson’s ratio of the 2195 Al-Li alloy were used, and their value was set to 71 GPa and 0.28, respectively. Stress contour is showed in Figure 6(d). According to Figures 6(a) and 6(d), the hole bottom is the stress concentration area, and the two cracks, A and B, initiate from that. The growth directions of the cracks, symbolized with thick arrows, were perpendicular to the loading direction, generally, and the crack paths were located in

![Figure 3: BSE micrographs of 2195-T8 alloy: (a) 500x magnification; (b) 1,000x magnification of the dotted frame in (a); (c) 3,000x magnification of the dotted frame in (b); (d) 5,000x magnification of the dotted frame in (c).](image-url)
the stress concentration band. The initiation of cracks A and B seemed to develop from a grain boundary of different grains, separately, in Figures 6(b) and 6(c), showing the magnified view of region I. The crack deflection, bifurcation, and stagnation were all observed by the observation along the propagation paths, though these phenomena took place only in GBs. Cracks A and B initiated and grew along GBs owing to the strength degradation of GBs affected by corrosion. The crack propagation paths were limited by the range of stress concentration band, which was vertical to axial loading. Notably, cracks A and B propagated along the single side of the hole, which did not conform with the final fatigue failure caused by the single main crack [3, 17–22]. Therefore, the invisible interconnection of cracks A and B should occur at the inner alloy. The details are discussed as follows.

Figure 7 shows the reflection of various regions signed in Figure 6, which were the typical fatigue crack characteristics along the crack paths from initiation to propagation. Region II (in Figure 6(b)) represents the initiation morphology of the crack A in Figures 7(a)–7(c). Irregular pits and a circular pit, displayed by arrows, existed in the site and could be considered as the source of the cracks. The crack initiated from those sites and grew along GB. The pits in Figures 7(a) and 7(b) are different from the pit shown in Figure 7(c). The formation mechanism is related to the discontinuous secondary precipitates, which were the same in Figure 4(a). The discontinuous precipitates were Cu-rich, leading to their worse electrochemical activity compared with the matrix [33]. They acted as a cathode and promoted the dissolution of the matrix during the corrosion process, which degraded their bonding strength with the matrix. Thus, they become weak sites along the GB after corrosion. Furthermore, the modulus of those hard and fragile discontinuous precipitates was incompatible with the matrix; hence, the deformation between those and the matrix showed the difference under a cyclic loading, which led to the worsening of the bonding strength [16]. Furthermore, the discontinuous precipitates hindered the dislocation motion, and those sites would evolve into local stress concentration sites as the cumulation of dislocation during the fatigue process. Therefore, the discontinuous precipitates could induce crack initiation or fall off from the matrix and leave irregular pits [7–9], as shown in Figures 7(a) and 7(b). Combining the influence of the corrosion and cyclic loading, the sites of discontinuous precipitates became crack initiation sites. Furthermore, GBs had higher energy with massive defects and weaker strength than the inner grains. From the EBSD results, several high-angle GBs existed in the alloy, suggesting that they had low corrosion resistance and were easily attacked during the corrosion [34]. The pit shown in Figure 7(c) proved this.

Figure 4: TEM micrographs of 2195-T8 alloy: (a) gain boundary region; (b), (c) Cu and Zr distribution in (a); (d) inner grain region; (e), (f) Cu and Zr distribution in (d), respectively.
Consequently, when the cracks initiated from the discontinuous precipitates in GBs, they would continue to propagate along the GB.

Figures 7(d)–7(f) display the magnified view of region III. In Figure 7(d), crack B passes through a chainlike pit and crosses certain grains near the pit, as indicated by arrows. The corrosion pits could be regarded as the secondary particles with zero modulus, which led their local strain rates to be larger than other sites under cyclic loading [10–12]. Adding to stress concentration effect, a crack tended to approach them. The chainlike pit, which has a large size made a strong dislocation motion and high stress in the grains around it. Besides, the crack grew to 0.8 mm, and its driving force increased. The above-said reasons made the crack cross several grains directly, and it indicated the role of the chainlike pits on the promotion of crack propagation. Figures 7(e) and 7(f) show the detailed crack deflection. The intergranular crack deflected upward for 33 μm and then to its initial propagation direction. In these two deflections, the crack met different grains. Intracrystalline high strength was higher than that of the corroded GBs, adding to the blockage of high-angle GBs for cracks; hence, the crack would only grow along the GB. Furthermore, when the crack direction was vertical to the loading direction, the crack could get large driving force with high stress intensity factor [35, 36]. This also contributed to the secondary deflection for crack propagation.

Region IV is shown in Figures 7(g) and 7(h). Figure 7(g) shows the crack bifurcation along the propagation direction. A recrystallized tiny grain is located and cracks appear on both sides of it. The bifurcated crack, symbolized with a dotted arrow, seemed to grow in reverse direction compared with the main crack direction. This phenomenon was attributed to the presence of the tiny grain boundary. From the EBSD analysis, this recrystallized tiny grain had both high-angle GB and large misorientation. They were susceptible to corrosion. Alternatively, under fatigue loading, their deformation was incompatible to other larger subgrains, thus leading to strain localization [15, 16]. Therefore, the tiny grain itself was the source of the microcrack initiation, which succeeded to explain the formation of the bifurcated crack. To a certain extent, the existence of recrystallized tiny grains decreased the corrosion resistance and degrades alloy strength. Furthermore, in Figure 6(b), the two sites of the crack bifurcation were also observed as shown in solid circle. For the blue one, the two microcracks were initiated besides the main crack, because the strong stress concentration could be induced near the crack tip. When the local stress
reached critical intergranular strength, intergranular cracks would initiate. For the purple one, the crack changed its initial propagation direction and grew along the upper GB, though a pit appears on the lower GB. Along the path, the arrest of the secondary crack might be caused by the interaction of the intergranular strength, GB misorientation, and stress magnitude. Thus, the main crack deflected towards the upper GB to propagate continuously [10, 16, 20].

3.3. Crack Propagation Analysis in 3D View. For the further investigation of the crack propagation behavior from the precorroded hole, computed tomography (CT) has been employed to reflect the fatigue crack in 3D view. As shown in Figure 8, the sample has been scanned from front to back (front view), from right to left (side view), and from bottom to top (vertical view). Along a scanning direction, the sample was divided over 500 scanning planes. By employing those planes, the sample was 3D reconstructed, as displayed in Figure 8(b).

Figure 9 shows the representative planes of the crack propagation morphology in front view along the normal direction and white lines display the outlines of pores in a specific plane, whereas the figures below in panels (a–e) displayed the relevant crack without considering the pores. In Figures 9(a) and 9(b), planes passed through the precorroded hole. When the cracks were within the projection surface of the precorroded hole, the cracks were vertical to the loading direction. But once they crossed the hole edge, they propagated at 45° from the loading direction, which was attributed to the maximum shear stress of the crack surfaces at 45° direction [35]. From that, the crack propagation direction was affected by the loading modes and crystallographic factors, and it achieved a balance between the two.

In Figures 9(c)–9(i), the planes located in the matrix showed crack propagation behavior in view of depth. The plane in (c) was near the alloy surface and the cracks are displayed discontinuous by the dotted circular lines. The crack discontinuity was caused by the high angle recrystallized grains. Their existence enhanced the difficulty of the crack growth and was one of main reasons for the multicrack initiation. Furthermore, the pores may contribute to the connection between the left crack and middle part shown by a yellow frame. In Figure 6, cracks A and B were not interconnected from the superficial observation. However, owing to the increase of the depth, crack B climbs up to the surface of the crack A, and they combined to form an intact and continuous main crack, as shown in Figures 9(d) and 9(e). When the planes were in deeper positions, noticeably, the crack length became shorter and merely located at the middle of the sample, indicated in Figures 9(f)–9(i). Owing to the distribution of the corrosion pits and defects at the alloy surface, the continuity and integrity of the alloy has been extremely affected. Under axial loading, the stress level at the exterior alloy surface was higher than that of the inner surface, which led to the cracks near the surface having larger driving force. Moreover, references [26, 37] reported that the fatigue crack surface could be equivalent to semieliptical shape during the whole fatigue process. Thus, the length of a crack in various positions was different under the combined effect of the stress level and shape evolution.

Figure 10 displays the typical side view of the cracks in various scanning planes, from right side to left side along
the transverse direction. The crack surfaces were perpendicular to the loading direction affected by the maximum normal stress; thus, the cracks would have large driving force to grow. Combined with Figure 6, the cracks were intergranular which are sensitive to misorientations among the grains. Thus, the crack would grow straight forward when there are no barriers in front of the crack tips. Nevertheless, according to Figure 10(c), the two crack surfaces in different heights are combined, and the higher one zigzags down to the flat surface. The higher crack surface was caused by a high-angle GB, which resisted the initial propagation direction; thus, the crack could only choose a twisted path to recover its initial direction [18, 19]. This also affected its driving force, which led to the arrest along the depth direction. From Figure 10(d), the cracks A and B, originally separated by a subgrain, tends to connect each other. The crack B would bypass the middle-grain to combine with A as shown in a circular line. Although the cracks A and B initiated from a

Figure 7: SEM micrographs of crack characteristics along paths of cracks A and B: (a–c) magnified view of region II in Figure 6(a); (d–f) magnified view of region III in Figure 6(a); (g, h) magnified view of region IV in Figure 6(a).

Figure 8: (a) Schematic diagram of CT scanning; (b) 3D reconstructed model.
GB of the same grain, their capacity for crack propagation was different when observed with respect to their growth depth. The differences were in the stress levels, precorrosion degree, and number of grains in different GBs, or other factors. Figures 10(g)–10(i) represent the bifurcation and merging process. A short secondary crack was above the main crack in (g), while it merged with the main crack in (h) as the crack propagated. This phenomenon illustrates that the main crack could connect to its adjacent secondary cracks owing to the large driving force, and the appearance of the secondary cracks also influenced the main crack propagation path. References [23–25] also confirmed that the secondary cracks could reduce the driving force of the main crack, and the tortuous crack path extended the crack length further to increase the fatigue life.

Figure 11 shows the evolution of crack surface in the vertical view along the rolling direction from the bottom to the top and also shows the influence of the pores in the matrix. Apparently, the pores, shown in red dotted frames, were located at the bottom of the artificial hole. According
to the distribution of the crack surfaces, displayed by yellow frames, the area size of the crack surfaces at the two sides of the hole was lower than that at the center of the hole. Furthermore, the crack surface area size was different at various planes, which indicated that the propagation of the main crack surface is tortuous. It reflected that the barrier from GBs, misorientations, and microstructural texture have made a strong influence on the crack propagation. According to Figures 11(g)–11(j), the area size of the crack surface increased gradually when the scanning plane got closer to crack A, and hence, the formation and propagation of the main crack were primarily dominated by crack A during the fatigue process.

According to the 3D view results of crack morphology, the porous region was distributed around the hole bottom and extended in a cylinder-like fashion along the loading direction. They seemed to promote the evolution and propagation of the crack surface displayed in all views. To verify that, the FEM results for the von Mises stress distribution along several traces on the hole section are displayed in Figure 12. Due to the irregular shape in the precorroded hole, the stress distribution in different directions is uneven.

Figure 10: Side views in various scanning planes along the transverse direction.
In order to consider the effects of all kinds of stress, von Mises stress was used instead of the stress in loading direction. Figure 12(b) displays the stress variation along the trace lines which are 0.9, 1.0, 1.5, and 2.0 mm from the alloy surface, separately. The stress increased from the sides and reaches the peak at the middle for every trace. When a trace was closer to the hole bottom, the stress peak would be displayed higher by the stress distribution along the 0.9 mm trace. Thus, the artificial hole and the corrosion pits would cause extreme stress distribution unevenly at different cross-sections located on various positions. Adding to the sandhill-like stress distribution along the traces, the 3D stress distribution tended to be a cylinder-like region. Nucleation and growth of the pores were always related with stress distribution.
Under the cyclic stress, the microplastic damage was distributed on GBs or grains with optimum misorientation, thus causing large stress near the bottom of the corroded hole. With damage accumulation, the micropores nucleate and grow. In this work, the loaded stress did not reach the yield stress; thus, the growth of micropores would be slow around the precorroded hole. Once micropores were nucleated, the matrix would be considered to possess more particles with zero modulus. Compared with the grains, the strain rate of the pores was higher, which promoted the increase of the size and proportion of the pores. As the fatigue process continues, the micropores might grow and connect to each other to cause strain localization, which promotes the size increase of the crack plane [3, 15]. Consequently, the clusters of pores formed the cylinder-like region contributed to the combination and propagation of the cracks under the cyclic loading conditions.

3.4. Fracture Analysis. Figure 13 shows the fracture surface of the sample with a precorroded hole. Figure 13(a) shows the overall fatigue fracture including a fatigue source region and propagation zone. Accordingly, the dotted crack source region was distributed at the bottom of the corroded hole. The propagation zone displayed radial river-like patterns on the fracture surface, and some slider secondary cracks were vertical to the crack growth direction [38, 39]. Those cracks seemed to be intergranular along the silver GBs. Furthermore, a step appears between the two crack surfaces, which were at different heights and they combined to each other with the crack growth. Thus, the multicracks involve

Figure 13: SEM images of the fracture surface: (a) overall view; (b, c) fatigue crack source zone; (d, e) typical features in fatigue crack propagation region; (f) final fracture zone.
the combination to form a single main crack according to the energy release criterion [15, 23]. Furthermore, a long GB, shown in dotted yellow lines, hindered the crack growth along the normal direction, which might change the initial growth tendency of the crack surface.

Figures 13(b) and 13(c) illustrate the crack source zone. The fatigue cracks originated from the defects at the surface and corrosion pits, since they both induce the stress concentration and large damage under cyclic loading. There are several crystallographic planes at different heights, which displayed river-like striations. As shown in thick arrows, owing to the effects of grain structures, misorientations, and textures, the paths of the striations in different grains were diverse [36, 37]. Figures 13(b) and 13(c) show the region of the fatigue crack propagation. There are multiple dimples and shallow fatigue striations that appear on the grains at the fracture surface. Furthermore, the secondary cracks were perpendicular to the crack growth direction and display the characteristic intergranular cracking. The most serious secondary cracking evolves into the delamination [3]. According to the figures, there were large cavities or high-angle grains among GBs, which significantly influenced the continuity of GBs. Thus, the dislocations were blocked by them and accumulate as the fatigue continues. From the fracture surface, the grains along the propagation directions were squashed, which led to prolonged GBs. GBs were more fragile than the inner grains; hence, GBs around the cavities and high-angle grains will be cracked when the accumulated stress reached the ultimate intergranular strength. The intergranular cracking continually induced delamination in the alloy, thus enhancing the little barriers along GBs. However, the role of the delamination in the Al-Li alloys is still being studied. Figure 13(f) represents the final fracture zone and shows the fraction modes mixed with the intergranular and quasi-cleavage. The delamination promoted the final fracture to a certain extent [40, 41].

3.5. Crack Bifurcation Simulation. To represent the crack propagation evolution restrained by the sliver barriers, as shown in Figure 13(a), the fracture mechanics could be used to solve it. Owing to the combination of the precorrosion and defect factors, the period of the crack initiation will be shortened. It was feasible to study the fatigue crack propagation by employing an equivalent initial flaw model. Based on the measurement and observation from the results of CT views and the fracture surface, an initial semielliptical crack surface, with 150 μm length, 45 μm depth, and 3° tilt angle, has been set and inserted into the 3D FEM model, as displayed in Figure 14. One end of the sample was fixed, and the other was applied constant axial fatigue loads whose stress ratio was 0.1. The element type and number were C3D8 and 35927, respectively. To improve calculation efficiency, the FEM model has been divided into submodels. A no-crack region was set to signify the barrier, as shown in the yellow region [42].

Paris law can reflect crack stable growth rates which was confirmed in [43]. However, for considering the effects of multiparameters, NASGRO formula, evolved from the Paris law, has been used to simulate the crack growth [43, 44]:

\[
\frac{da}{dN} = C \left( \frac{1 - f}{1 - R} \right)^n \left( 1 - \frac{\Delta K_a}{\Delta K_c} \right)^p \left( 1 - \frac{K_{max}}{K_c} \right)^q,
\]

Table 2: Comparison of test and simulation life.

| Sample number | Test life | Simulation life | Error (%) |
|---------------|-----------|-----------------|-----------|
| #1            | 33590     | 30234           | 9.9       |
| #2            | 34541     | 30234           | 12.4      |
| #3            | 28632     | 30234           | 5.6       |

Figure 14: FEM modeling and crack surface evolution.
where $C$, $n$, $p$, and $q$ are constants and $f$ is the shape function. Further, $R$ is stress ratio, $\Delta K$ is the stress intensity factor range, $K_c$ is the fracture toughness, and $\Delta K_{th}$ is the threshold value.

The crack front propagation length $\Delta a_i$ can be calculated by specifying the load cycles step $\Delta N$ [42] that can be expressed as

$$\Delta a_i = \Delta a_{\text{median}} \left[ \frac{da/dN_i(\Delta K_i, R_i \ldots)}{da/dN_{\text{median}}(\Delta K_{\text{median}}, R_{\text{median}} \ldots)} \right], \quad (2)$$

where $\Delta a_i$ and $\Delta a_{\text{median}}$ are the crack front propagation extensions at the $i$th discretized point and the median point.

After 125 crack growth steps, the fatigue crack growth simulation was completed. As shown in Table 2, the errors between the simulation and test life results were up to 12.4%, which showed the accuracy of the simulation. The evolution of the crack surface is displayed in Figure 14. The crack grew faster along the length direction than that in the depth direction. When the crack front met the barrier, the restrained part stopped growing, and the original front was separated into two fronts that made the crack grow towards the two sides. This is presented in step 106 and step 110.

Figure 15 reflects the distribution of $K_I$ along the crack fronts during the whole fatigue process. Owing to the mode I loading condition, $K_I$ is the dominant parameter, and thus, the results of $K_{II}$ and $K_{III}$ are not displayed. Before the crack surface met the barrier, which is shown by a red dotted line, the change of $K_I$ could be divided into two stages. When the crack front was far from the barrier, $K_I$ values along the front were similar, thus illustrating similar growth rates in all directions. However, once the front was close to the barrier, over 100 steps, the $K_I$ values near the sample surface were much larger than those in the deep positions, which revealed that the driving force of the crack growth along the length direction was larger. That result could explain the discrimination of the crack growth in various directions. When the crack was hindered by the barrier, which evolved into two fronts, the $K_I$ values were huge along the fronts. That phenomenon was caused by the redistribution of the fracture energy, which made the two fronts share the initial energy in the same magnitude. Furthermore, the $K_I$ values would reach over 2500 MPa$\cdot$mm$^{1/2}$ near the surface at step 111, which led to the fast fracture after that step.

4. Conclusions

This paper presented a systematic investigation of the fatigue crack propagation behavior of the 2195-T8 Al-Li alloys with a precorroded hole, through experiments and simulation. This investigation deepened the understanding of fracture mechanisms and promoted the application of Al-Li alloys. The results are summarized as follows.

1. Precorrosion weakened the intergranular strength and promoted the formation of corrosion pits at the bottom of the artificial hole. Corrosion pits and corroded discontinuous precipitates were the primary fatigue crack nucleation sites. The attacked high-angle GBs made the crack propagation primarily intergranular and caused deflection and bifurcation.

2. Corrosion pits increased the stress concentration around the artificial hole and promoted the cracks to grow faster near the corroded surface than that inside the alloy. Superposed stress stimulated the evolution of the cylinder-like porous region at the hole bottom, affecting the fatigue crack increments and interconnection.

3. Delamination and secondary cracks were aligned vertical to the crack growth direction and had significant effects on the evolution of the single main crack front. Hindered by the barrier, multicrock fronts would form and induce higher $K_I$ to grow fast, thus causing the final fracture by turning to the sides of the original direction of propagation.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.
Conflicts of Interest
The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Authors’ Contributions
Dejun Liu and Gan Tian are responsible for the investigation, data curation, visualization, writing—original draft, and modification and Yulong Li, Guofeng Jin, and Wei Zhang for the supervision, conceptualization, methodology, and funding acquisition.

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