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An experimental methodology to quantify the resistance of grain boundaries to fatigue crack growth in an AA2024 T351 Al-Cu Alloy

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ABSTRACT

The growth behaviors of the short fatigue cracks initiated from the micro-notches, $30 \times 15\ \mu m^2$, fabricated using FIB in some coarse grains, were investigated across grain boundaries in an AA2024-T351 Al-Cu-Mg alloy in four-point bend fatigue at 20 Hz, R = 0.1, and room temperature in air. The resistance to short crack growth at the grain boundary that the crack crossed was quantified from the measured decrease in growth rate at the boundary. It was found that the decrease in crack growth rate followed a linear relationship with the twist component (α) of crack deflection at the boundary, as revealed in EBSD and FIB experiments. The resistance of grain boundaries to short crack growth was revealed to a Weibull-type function of α, which quantitatively verified Zhai's minimum α criteria for short fatigue crack growth across grain boundaries in planar slip alloys.

1. Introduction

Thorough and quantitative understanding of the early stage of fatigue crack growth is of scientific and engineering significance, since crack early growth could take over 90% the total life of an engineering alloy [1]. Unlike a long crack, growth of a short fatigue crack is profoundly sensitive to the local microstructure [2], such as grain boundaries (GBs), phase boundaries, particles and texture, etc., in alloys. Among these microstructural features, GBs have been recognized as the major barrier to short fatigue crack growth [3–7], which accounts for the marked variation of the short fatigue crack growth rate measured in most engineering alloys [2,8–10]. Over the past decades, numerous efforts have been made in attempts to quantitatively understand the behaviors of short fatigue crack growth through GBs. Chan and Lankford et al. [11,12] proposed a crack tip strain model to predict the growth behaviors of short fatigue cracks by suggesting that the maximum plastic strain amplitude in the vicinity of the short crack tip was a power law function of the driving force (ΔK) for crack growth. As a result, crack growth deceleration at a grain boundary could be explained to be attributed to the strain amplitude ahead of the crack tip being reduced due to blocking of the slip band ahead of the crack tip by the boundary. Based on this model, Li [13] later also took into account the effect of the secondary slip bands on the crack tip displacement, and assumed that the crack tip opening displacement was a result of the crack tip slip on both the primary and secondary slip bands. In order to quantify the effect of slip blocking by a GB on short crack growth, several models [14,15–18] were subsequently developed, mostly based on Bilby-Cottrell-Swinden's (BCS) theory [19] which describes the crack growth and the plastic deformation zone in the vicinity of the crack tip as an array of continuously distributed immobile dislocations. However, these models could not directly account for the effects of the local texture and its change across a GB on crack growth though the GB, though De los Rio et al. [16], incorporated the Sachs factor averaged over all the grains along the crack front into their model for short crack growth simulation. Ravichandran and Li [20] studied the effects of variation in elastic moduli of the grains covered by a short crack on the stress intensity factor (SIF-K) along the crack front. Using a weight function technique and numerical analysis, K was calculated and found to vary along the crack front due to the variation of grain orientation. Although they were able to explain the effect of grain orientation variation on the shape of a surface short fatigue crack, the influence of GBs on crack growth was not considered quantitatively. In a near isotropic material such as Al, the anisotropic effect of elastic modulus due to grain orientation variation might be neglectable and GBs become the dominant barrier to crack growth. Their effects have to be considered in

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order to establish a true microstructure-based model for simulating short crack growth in engineering alloys.

A crack tip displacement (CTD) parameter, obtained based on edge dislocations slipping out of the crack tip, was also used to calculate short fatigue crack growth rates [17,18]. The effect of grain orientation was simulated on the crack growth in an Al-Li alloy using this model [17]. The CTD range has recently been used in simulation of short crack growth from either pores or particles on the sample surface [21–23]. Although the effects of grain size and orientation were discussed in the model, their contribution to the crack growth rate could not be incorporated in calculation of the crack growth rate [23]. Li et al. [24] used an effective concentration factor, $k_g = \sqrt{k_\sigma k_\varepsilon}$, where $k_\sigma and k_\varepsilon$ are stress and strain concentration factors, respectively, at a defect or crack tip to calculate short crack growth, since it was assumed that the crack followed the path where k_g was the maximum. The crack growth rate was then proposed to be, $da/dN = Ck_g^m$, where C and m are materials constants. However, this method was unable to account for crack retardation at GBs, since it could only predict that the crack should grow faster at the GBs, as k_g was much higher at the GBs than inside the grain. McDowell et. al. [25–27] have recently suggested a microstructure-sensitive multistage model which quantifies the fatigue life in three separate stages, namely, crack incubation, short and long crack growth. The life of crack incubation is estimated with a microscopic scale Manson-Coffin equation at a pore or a particle, whereas the short crack growth is calculated based on a crack tip displacement model [27].

All the above mentioned models are, however, 2 dimensional in nature, thereby unable to deal with the 3 dimensional effects of microstructure on short crack growth. For an example, a GB can still interact with a crack after the crack passes through the GB on the surface [28]. With electron backscatter diffraction (EBSD) and acoustic microscopic techniques, Zhai et al. [28–30] have identified that the twist component of short fatigue crack plane deflection at a GB is the key factor controlling the crack growth behaviors across the GB in Al-Li alloys, and proposed a minimum twist criterion for the selection of the short crack path across the GB, i.e., the crack preferably follows the slip plane that has the minimum twist angle with the crack plane at the GB in planar slip alloys. This has revealed the physical mechanism for the resistance of a GB to short fatigue crack growth in these alloys, since extra energy is required to form fracture steps to accommodate the crack plane twist at the GB [28]. The GB resistance depends on the crystallographic geometry of the GB which determines the deflection of the crack plane at the GB. The minimum twist criterion has also been verified in Ni-based superalloys through revealing the 3-D geometric relationship between short crack planes and GBs by cross-sectioning the interacting crack and GB using focused ion beam (FIB) and analyzing the crack plane orientation using EBSD by Schaef and Marx et al. [31–33]. Using the combined diffraction and phase-contract X-ray tomography, Ludwig et al. [34–36] were able to obtain an in-situ 3-D landscape of the short crack growth through GBs in a metastable beta titanium alloy. Their results also supported Zhai's minimum twist criterion for short crack growth across GBs [28]. However, it is still desirable to establish a quantitative relationship between the GB crystallographic geometry and the resistance to crack growth, in order to develop a true microstructure-based model to simulate fatigue crack growth in 3-D.

In this research work, an experimental method was developed to quantify the resistance of a GB to short fatigue crack growth in AA2024 T351 Al alloys. It was found that the resistance was a Weibull-type function of crack plane twist at the GB. This could pave a way to simulating short fatigue crack growth in 3-D, based on the microstructure and texture in the alloys.

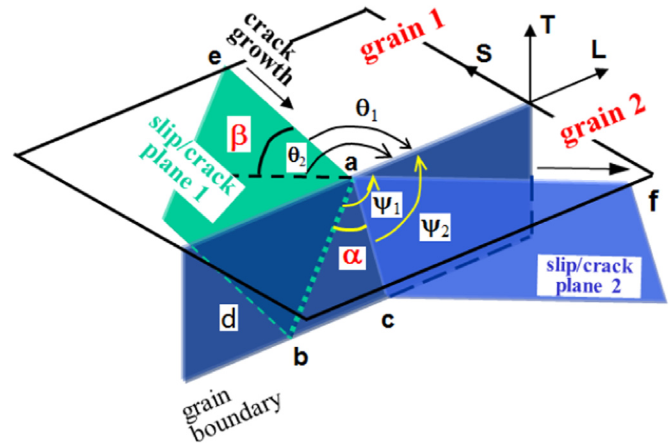

Fig. 1. Schematic diagram showing a crystallographic mechanism for crack growth along slip plane 1 in grain 1 onto slip plane 2 in grain 2. The crack growth across the GB is controlled by α and β (after [18]).

2. Zhai's 3-D crystallographic model

Fig. 1 is the 3-D crystallographic model proposed by Zhai et al. for short crack growth across a GB (for details, see references [28,30]). In this model, the geometry of the GB is described with twist (α) and tilt (β) components of crack deflection at the GB, rather than the commonly used GB misorientaion. $\boldsymbol{\alpha} = |\psi_1 - \psi_2|$, defined as the angle between the two intersecting lines (*ab* and *ac*) of the two favored slip planes on the GB plane, represents the major resistance to crack propagation from one grain to another because the wedge shaped area (*abc*) described by α on the GB plane has to be fractured for the crack to pass through the GB. The larger the α, the higher the resistance from the GB. The angle between the intersection lines of the two favored slip planes on the sample surface is $\boldsymbol{\beta} = |\theta_2 - \theta_1|$ which can also contribute to the total resistance of the GB to crack growth by reducing the driving force at the crack tip. For a short crack to cross the GB from grain 1 into grain 2, it has to overcome the resistance from the GB. As a result, at the GB, the crack will select the slip plane that offers the minimum α angle, i.e. minimum resistance, among the four slip planes in grain 2, since the crack preferably follows a slip plane within a grain. This model is able to explain the 3-D effects of GBs on short fatigue crack growth in planar slip alloys, such as crack deflection and retardation at the GBs, since a GB could still drag the crack even after the crack passes the GB on surface. Among **α**, β and Schmidt factor, α is the most dominant factor controlling the short crack growth [28,30]. Although the resistance of a GB to short crack growth is defined by the crystallographic model, it still needs to be quantified for use in simulating short crack growth in 3-D.

3. Experimental details

3.1. Material

The material used in the present work was a 7 mm thick AA2024-T351 Al alloy plate (its chemical composition is listed in Table 1) in which grains were pancake shaped (361 × 97 × 37 μm^3

Table 1
Chemical composition of AA2024-T351 (wt%).

Al	Cu	Mg	Fe	Mn	Si	Cr	Ti	Zn
Balance	4.1	1.4	0.45	0.43	0.43	< 0.1	< 0.15	< 0.25