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A 3-D view on the mechanisms of short fatigue cracks interacting with grain boundaries

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Abstract

Short fatigue cracks propagate with a fluctuating crack growth rate while interacting with microstructural barriers. Although some models of the interaction exist, it is not clear what determines the strength of the resistance of a grain boundary to crack propagation. Therefore we developed a method for artificial crack initiation which uses a focused ion beam (FIB) to nucleate cracks crystallographically on single slip planes identical to natural stage I cracks. The crack parameters as well as the grain boundary parameters can be varied independently for systematic experiments. For the first time the crack path through a grain boundary is shown in 3-D by FIB tomography. This enables the interaction between microcracks and grain boundaries to be observed in 3-D with a high spatial resolution. It is not only the inclination angle between the active slip systems, but also the inclination angle of the grain boundary, which determines the strength of these microstructural barriers.

Keywords: Fatigue crack growth; Short cracks; Grain boundaries; Crack arrest; FIB tomography

1. Introduction

Under fatigue conditions crack growth usually starts with short fatigue crack propagation in stage I. The growth of short cracks cannot be described in the same way as that of long cracks where the crack propagation per cycle, da/dN, is a function of the effective cyclic stress intensity factor ΔK_{eff} as written in the Paris law [1]:

$$\frac{da}{dN} = D \cdot \Delta K_{eff}^{m}, \tag{1}$$

where D and m are fitting parameters and reflect in some way the average influence of the microstructure. ΔK_{eff} accounts for the ratio between the maximum and minimum load considering crack closure effects.

Nevertheless applying Eq. (1) for short cracks results in highly fluctuating crack growth data since the cracks grow on different slip planes in each grain and strongly interact with the adjacent grain boundaries [2–4]. This paper describes how the mechanisms of the interaction were elucidated with a very sophisticated experimental setup using a focused ion beam (FIB) microscope. In order to understand the details, the models which describe the interaction will be briefly described.

In stage I, cracks grow by the emission of dislocations from the crack tip on the slip plane which is identical to the growth plane (Fig. 1). The local stress intensity operating in this crack propagation mode has already been described by Bilby, Cottrell and Swinden—the so-called BCS model [5]. The crack growth per cycle is determined by the crack tip sliding displacement (CTSD) which results from the number of dislocations emitted at the applied load. This can be expressed in similar fashion to the Paris law:

$$\frac{da}{dN} = C \cdot \Delta \mathrm{CTSD}^{n}. \tag{2}$$

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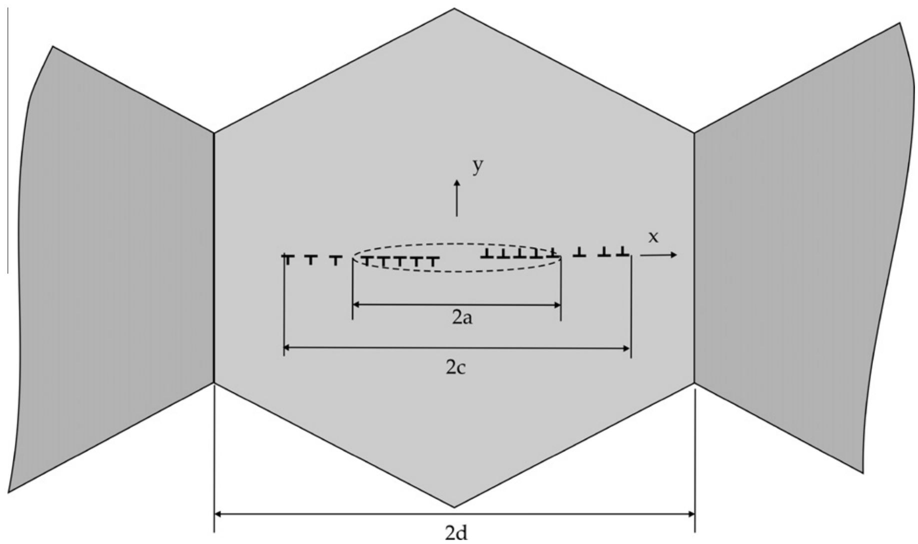

Fig. 1. BCS model of a crack propagating in stage I inside a single grain.

Here the CTSD can be calculated from the dislocation model, C and n are material parameters corresponding to D and m from Eq. (1). For a single crystal (SX) all parameters are now well defined.

However, the model can only be used as long as a crack is lying completely, including its plastic zone, inside a single grain. When the plastic zone in front of the crack tip comes into contact with a grain boundary, the dislocations are blocked, which alters the local stress state and hence the crack tip opening displacement. This was observed, for instance, by Zhang and Edwards in an aluminium alloy [6–8]. They used electron channelling contrast imaging (ECCI) to investigate the spreading of the plastic zone ahead of microstructurally short cracks. They found a decrease in the plastic zone size, or even a complete blocking, while crossing a grain boundary. The strength of the barrier was observed to be stronger for larger misorientations between adjacent grains. This is described quantitatively by the models of Tanaka and Navarro and De Los Rios [9,10] which are based on the observation that short cracks propagate crystallographically into the neighbouring grains as long as they are microstructurally short [11]. A good overview is given in Ref. [12].

There are several parameters which determine the strength of the interaction between cracks and grain boundaries. These can be categorized as crack parameters, e.g. crack length, distance to the boundary and applied load or stress intensity factor, on the one hand, and grain boundary parameters, e.g. boundary type and orientation, on the other. The latter measure the resistance of a grain boundary to crack propagation. The crack parameters strongly influence the stress field at the crack tip. Most important is the crack length prior to the distance between crack tip and grain boundary. At a constant applied load, the crack length acts as driving force of the dislocation emission and motion. It is obvious that longer cracks have a larger CTSD [13]. However, the CTSD is additionally affected by the distance between crack tip and grain boundary: the plastic zone in front of the crack must at least interact with the boundary to show an influence of the microstructure on crack propagation. In this case the boundary is considered as a simple barrier neglecting additional effects arising from elastic and/or plastic incompatibilities near the boundary. A typical grain boundary parameter is the orientation difference of the adjacent grains coupled with the orientation difference of the potential slip systems. Another parameter which is often neglected in most descriptions and models is the inclination angle between surface and grain boundary. So far nearly all models describe grain boundaries oriented perpendicular to the surface. In real materials this is often not the case.

In the literature there are two established models to describe the resistance of the grain boundary to crack propagation. However, they consider the grain boundary parameters in different ways. The first model of Tanaka, further developed by Navarro and de Los Rios, describes the resistance of a boundary to crack propagation only by the orientation of the neighbouring grain. This has consequences for the suitable slip planes in the second grain and the dislocation emission on the appropriate slip systems. The inclination angle of the grain boundary is not further considered in this model. Although the Tanaka model includes different grain orientations it is one-dimensional. It was further developed for two dimensions using the boundary element method [14], which enables it to take at least the 2-D orientations of slip systems in adjacent grains as well as crack closure effects into account. By adapting the model to experimental data, a good correlation was observed. However, experimental data are scarce and 3-D aspects have not been considered. A further improvement has been done by the extension of the model to 3-D. Numerical studies [15,16] have shown the potential of 3-D calculations using hyper-singular integral equations based on the theory of Hills et al. [17]. For experimental