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Dislocation-based modelling of the growth of a microstructurally short crack by single shear due to fatigue loading

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Abstract

The growth of a short edge crack, located within one grain in a bcc crystal and subjected to cyclic loading, is modelled using a dislocation formulation. The external boundary including the crack itself is built from dislocation dipole elements and the plasticity is represented by discrete dislocations moving along preferred slip planes. The crack propagates by single shear due to emission and annihilation of dislocations, leading to a zigzag shaped crack path. The method allows simulation of the crack growth path, crack shape and the development of the plastic zone in detail. It is shown that the grain orientation, orientation of the activated slip systems, the load and the initial crack configuration all together determines the exact path.

Keywords: Short crack; Discrete dislocation; Dislocation dipole; Single shear; Stage I fatigue

1. Introduction

The part of life that a component spends in the Paris' regime before failure can be estimated from Paris' law [1]. It assumes that small scale yielding prevails and that linear elastic fracture mechanics applies. However, for short cracks, when the plastic zone size is significant in comparison with the crack length, linear elastic fracture mechanics cannot be applied and the crack growth is more difficult to quantify. The growth of such a short crack is driven by shear. The growth is strongly influenced by the microstructure of the material and the crack advance is under dominance of local plasticity. For advanced materials a large portion of the total component life might already be spent during propagation of short cracks, before the Paris' regime is entered, and it is therefore of importance to be able to model the behaviour of short cracks to improve fatigue life estimates.

It is well accepted that fatigue propagation of microstructurally short cracks is caused by localized plastic shear deformation close to the crack tip and numerous experimental evidence show that such cracks grow by a single shear mechanism. Suresh [2] and Andersson and Persson [3] have performed in situ microscopic studies that show that cracks grow in a zigzag pattern, following different slip planes. The same observations were made by Jono et al. [4] for a short crack in a body centred cubic (bcc) silicon iron. Zhang [5] has studied short fatigue cracks experimentally and proposed a shear band decohesion model for short cracks subjected to fatigue loading, in which the crack grows along specific shear bands, resulting in a zigzag shaped crack.

For very low growth rates, in the order of a few Burgers vectors per cycle, it is important to account for the influence of discrete dislocations. The discrete dislocation formulation by Riemelmoser et al. [6] and Riemelmoser and Pippan [7] models the interaction between a half-infinite mode I crack and individual dislocations. Bjerkén and Melin [8,9] modelled mode I fatigue growth within one grain of bcc iron of a microstructurally short crack based on a dislocation formulation where crack growth resulted from dislocation emission from the crack tip. Also the influence on the crack growth rate from grain boundaries was investigated. The crack deflection in bcc iron was modelled by Uematsu et al. [10] by superposition of the stress fields

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from individual dislocations and from finite element calculations of the crack stress field. In the model dislocations are assumed to nucleate when the resolved shear stress exceeds a critical value for a certain amount of time. Navarro and de los Rios [11] have developed a model that with one single equation describes the fatigue crack growth rate of short cracks including the influence from grain boundaries. They have also conducted experiments on medium carbon steel and good predictions of actual experimental lifes were achieved.

The approach in this paper is a combination of modelling plasticity by discrete dislocation movements along certain slip planes and describing the external boundary, including the crack itself, by dislocation dipole elements in the spirit of a boundary element approach, cf. [8,9]. The bulk material itself is considered isotropic and linear elastic. The method is used to study the quasi-static growth of a short edge crack, located within one grain and extending by single shear due to cyclic loading under plane strain conditions. The crack path is dependence of crystal orientations in a bcc structure and crack shapes are studied and the emergence of the plastic zone is followed in detail.

2. Problem formulation

Quasi-static growth due to varying load of a short edge crack in bcc iron is investigated under plane strain conditions. The crack is located within one grain and has an initial length a_0. Originally, the crack is straight and inclined an angle α to the normal of the edge of a semi-infinite body, cf. Fig. 1. The angle α coincides with the angle defining a preferred slip direction. A global Cartesian coordinate system (x,y), with origin at the position of the crack mouth, is introduced according to Fig. 1. In front of the crack, and parallel to the free edge, a grain boundary is present at a distance l_{GB} from the original crack tip.

An external load, σ_{yy}^{∞}, applied parallel to the edge, is varied between a maximum value, $\sigma_{yy\,\max}^{\infty}$, and a minimum value, $\sigma_{yy\,\min}^{\infty}$, cf. Fig. 1. The crack is assumed to grow through single shear along preferred slip planes as a result of dislocation nucleation. The preferred slip planes are separated an angle β within the grain, cf. Fig. 1, where the slip directions are indicated by dashed lines.

As the load is varied, the resolved shear stress is calculated at each load level at all slip planes along which dislocation nucleation and dislocation movements occur or can be expected. Such slip planes either emanate from the crack tip as illustrated in Fig. 1, or from corner points of the crack if the crack growth direction has changed so that a zigzag path is formed. As the resolved shear stress exceeds the nucleation shear stress, τ_{nuc}, at a distance r_{nuc} from the crack tip or corner point along a slip plane, edge dislocation pairs are assumed to be nucleated. The crack growth process is driven by this nucleation of dislocation pairs along preferred slip planes, followed by dislocation annihilation during external load reversal under the assumption that no healing of the crack surfaces occurs.

The material characteristics in this study are those of pure iron, which has a bcc crystal structure at room temperature. In a bcc material, full slip occurs in closed packed $\langle 1\ 1\ 1 \rangle$ directions, cf. Hull and Bacon [12]. Thus slip takes place on the close-packed $(1\ 1\ 0)$ plane in the $[\bar{1}\ 1\ 1]$ and $[1\ \bar{1}\ 1]$ directions, cf. Fig. 2, and the crack front lies in the $(1\ \bar{1}\ 0)$ plane. This results in two possible crack extension directions. Either the crack lies in the $[\bar{1}\ 1\ 1]$ direction, cf. Fig. 2.1, or in the $[1\ \bar{1}\ 1]$ direction, cf. Fig. 2.2. The resulting two dimensional equivalent slip systems are found by viewing the $(1\ 1\ 0)$ plane, cf. Fig. 2.3 and Fig. 2.4. Two possible angles, β, between the active slip directions, $\beta = 70.6°$ and $109.4°$, thus emerge.

The material is assumed to be isotropic with the anisotropy resulting from the crystal structure notable through the preferred slip plane directions, only. Thus the linear elastic material parameters used for bcc iron are the shear modulus $\mu = 80$ GPa and the Poisson's ratio $\nu = 0.3$, cf. eg. Askeland [13], giving the Kolosov constant $\kappa = 3 - 4\nu = 1.8$. The size of the Burgers vector

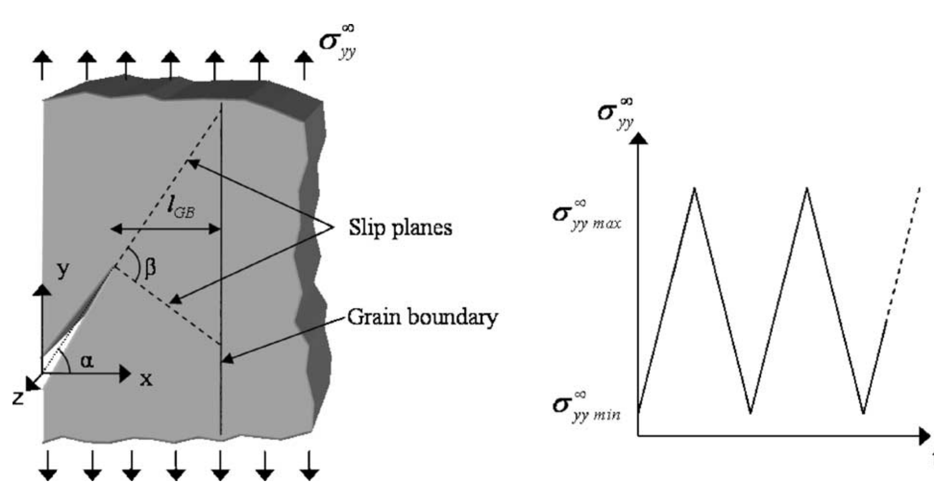

Fig. 1. Geometry of the short edge crack and external load cycles.