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Crack growth rates for short fatigue cracks simulated using a discrete dislocation technique

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ABSTRACT

The influence on the crack growth rate for a short edge crack under fatigue loading due to changes in crack length, grain size, load range and grain boundary configuration, is investigated under quasi-static and plane strain conditions. The geometry is modelled by distributed dislocation dipole elements in a boundary element method approach and the plasticity is described by discrete dislocations. The crack is assumed to grow due to nucleation, glide and annihilation of dislocations along slip planes in the material in a single shear mechanism. The results of the investigation are compared to typical growth rates for long cracks and it is found that the increase in growth rate due to a prescribed stress intensity factor range was much less pronounced as compared to what holds for long cracks.

1. Introduction

When a component is subjected to a fatigue load the damage process can, generally, be divided into three major phases: initiation, propagation of a short crack, and subsequent propagation of a long crack. Most conventional fatigue life estimate methods do not consider the different phases of the crack development. The so-called total life approaches are based on *S–N* curves developed by Wöhler already in 1860, estimating the fatigue life as a function of stress. A different approach is to take into account the growth behaviour of long cracks. The most well known is the work by Paris and Erdogan [1], using the concept of linear elastic fracture mechanics to evaluate the crack growth rate for long cracks. They found a correlation between the crack growth rate and the stress intensity factor range in the well-known Paris' law. However, in some cases, a large part of a components life is spent in the first two phases, i.e. crack initiation and propagation of a short crack. These primary phases of crack extension are therefore important to evaluate and quantify.

When a crack is shorter than a few grains the crack is said to be short and the concept of similitude used to formulate Paris' law do not apply. It was found by Pearson [2], among others, that a short crack can grow at high rates at low stress ranges, below the threshold value for long crack propagation. The difference in growth rate can be explained by the differences in growth mechanisms. Long cracks, called stage II cracks, typically grow in a duplex slip mechanism perpendicular to the applied load direction whereas short cracks, stage I, grow in a single shear mechanism along specific slip plane directions, cf. Suresh [3]. A number of experimental evidence for such growth of short cracks can be found. Sugeta et al. [4] have shown that short zigzag shaped cracks develop in silicon iron in the low K region, with K denoting the stress intensity factor. Also Zhang [5] have found that cracks grow along shear bands, created in front of the crack tip in an experimental study of ultra-fine grain sized aluminium. Düber et al. [6] have performed fatigue experiments on duplex steels and have, by following the paths of short cracks, been able to find the transition of single to double slip mechanisms in the material.

In an experimental study by Ohr [7] it was found that, during the early stages of crack propagation, linear arrays of dislocations were formed in front of the crack tip and that slip systems were activated at the crack tip. It was also found that some of the crack tip generated dislocations returned to the crack tip and annihilated during unloading. Therefore, in order to accurately simulate growth rates in the order of a few Burgers vectors only, it is important to take the generation and movement of discrete dislocations into account. Riemelmoser et al. [8,9] developed a discrete dislocation model to quantify the growth of a long mode I crack. The model was used to study the cyclic crack tip plasticity, influence of material parameters, crack length and loading conditions. A similar model was also developed by Bjerkén and Melin [10] to study the propagation of a short mode I edge crack in the neighbourhood of a grain boundary. A somewhat similar approach was employed by Künkler et al. [11], in which both the crack and the plasticity along the slip planes were modelled using distributed dislocations in a

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boundary element approach. The model was used to simulate the transition from a stage I to a stage II crack in duplex steel. Also an experimental study was made for comparison.

The presence of a grain boundary in the neighbourhood of a short crack will have a significant influence on the crack growth rates. There are several different mechanisms for dislocation slip propagation movements across grain boundaries. In a study by Shen et al. [12], a number of different mechanisms were identified and it was found that a grain boundary could act both as dislocation obstacle and as a source for dislocation generation. Bjerkén and Melin [13] have performed and investigation on how a grain boundary affects the growth behaviour of a short edge crack. In the model, it is assumed that the grain boundary acts as an obstacle, which the dislocations cannot pass. When the stresses in the neighbouring grain gets sufficiently high, nucleation of dislocations will occur in that grain, resulting in a spread of the plasticity. Another approach has been used by Doquet [14], assuming that if the stress at the leading dislocation in the pile up is sufficiently high it will enter the boundary and a new one will glide into the next grain on a somewhat misoriented plane. Similar observations have also been found by [7].

In this paper the growth of a microstructurally short edge crack, located within one grain in a body centred cubic (bcc) material subjected to fatigue loading, is simulated. The crack is assumed to grow in a single shear mechanism due to nucleation, glide and annihilation of discrete dislocations, with the crack tip being the only source of dislocation nucleation. The modelling is based on a dislocation formulation, were both the geometry and the plasticity are described by discrete dislocations. To model the geometry dislocation dipole elements have been used in a boundary element approach, and to model the plasticity and the resulting plastic deformation, discrete dislocations, located along specific slip planes in the material, have been used. The aim of this paper is to evaluate how the crack growth rate changes due to crack length, grain size, load range and grain boundary configuration. In this paper, also two different possibilities of modelling the grain boundary are considered.

2. Crack geometry

The crack growth of a microstructurally short edge crack located within one grain in a bcc structure, subjected to fatigue loading, have been investigated under plane strain and quasi-static conditions. The crack is assumed to grow in a single shear mechanism due to nucleation, glide and annihilation of discrete dislocations along slip planes in the material. In this study, it is assumed that one slip plane only, emanating from the crack tip and with direction coinciding with the crack direction, is active. This restriction, only using one slip system, is chosen in order to ensure that the crack will remain straight and not grow in a zigzag pattern on alternating slip planes as obtained in general cases, cf. Hansson and Melin [15,16]. From these earlier studies by Hansson and Melin it was found that changes in growth direction strongly influenced the growth rate and, in order to avoid large variations in growth rate, corresponding to changes in growth direction, only one possible slip plane is considered in this study. The initial crack, of length a, inclined an angle θ_1 to the normal of the free edge, is located within a semi-infinite body, cf. Fig. 1. The external fatigue load, σ_{yy}^{∞}, is applied at infinity, parallel to the free edge and is varied between a maximum value, σ_{yymax}^{∞}, and a minimum value, σ_{yymin}^{∞}. In Fig. 1, $\perp$ denotes the position of an edge dislocation.

In this study, two neighbouring grains are considered, where the grain boundary between the two grains is chosen to be perpendicular to the slip plane in the first grain, cf. Fig. 1. This choice is made in order to get an easy interpretation and implementation

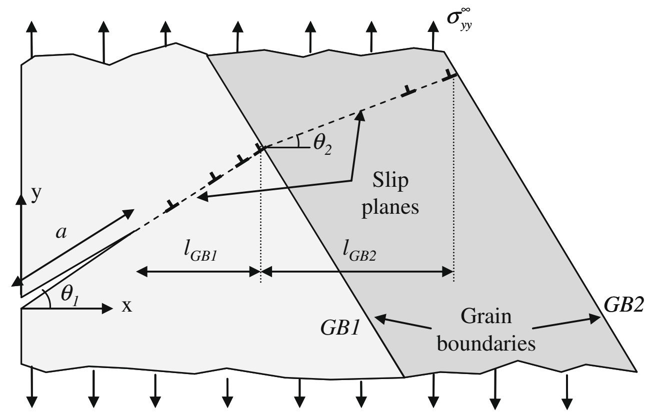

Fig. 1. Geometry and loading conditions of the short edge crack together with the spread of the plastic zone within two grains. $\perp$ denotes the position of an edge dislocation.

Table 1
Material properties and initial geometrical parameters.

Shear modulus, μ	80 GPa
Poisson's ratio, ν	0.3
Burgers vector, b	0.25 nm
Lattice resistance, τ_{crit}	40 MPa
Initial crack length, a_0	10,000–80,000b
Distance to grain boundary, l_{GB}	2000–15,000b
Crack angle, θ_1	45°
τ_{nuc}	1.59 GPa
r_{nuc}	6b

of the low angle grain boundary introduced in Section 3.5. The grain boundary is located at a distance l_{GB1} in front of the original crack tip position, measured as the distance between the crack tip and the point of intersection between the slip plane and the grain boundary in the global x-direction. A second grain boundary is also introduced at the remote end of the second grain, cf. Fig. 1, in order to ensure that the dislocations do not travel through the entire structure due to high applied loading. This grain boundary is, for simplicity of the model, chosen to be parallel to the first grain boundary and placed a distance l_{GB2}, measured in the global x-direction, away from the first boundary, cf. Fig. 1.

2.1. Initial conditions

The material in this study is pure iron with a bcc crystal structure and is assumed to be linear elastic with the emerging plasticity described by discrete dislocations located along specific slip planes in the material. The material parameters at room temperature are shown in Table 1, cf. Askeland [17], together with the geometrical data for the initial edge crack defined in Fig. 1. Also the nucleation stress, τ_{nuc}, and the distance between two newly nucleated dislocations, r_{nuc}, is given in the Table, both quantities defined later on in Section 3.3.

3. Boundary element approach

The modelling in this paper rests solely on dislocation formulations, were both the geometry of the short edge crack and its surroundings as well as the plasticity are described by dislocations in a boundary element approach. Only plane problems are addressed, and, therefore, only edge dislocations are needed in the formulation.