A NUMERICAL APPROACH TO THE FATIGUE CRACK GROWTH THRESHOLD

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ABSTRACT

The crack growth rates of 7075-T651 aluminum alloy found in humid air, technically purified nitrogen and a fine vacuum are significantly different near threshold. The average crack growth rate da/dN is theoretical explained as a micromechanical quantity multiplied by a probability. The micromechanical quantity has the dimension of a length. In the case of corrosive environments, e.g. air or nitrogen (with traces of water vapor and oxygen) it is the thickness of the oxide film. In the case of vacuum, it is the typical distance between cyclic slip planes. The da/dN versus ΔK_{eff} curve represents the probability per cycle, that the crack front is moving one step forward. A multi-body model developed by Masing for the stress-strain curve is adapted to the threshold in fatigue crack growth. In our model, the material elements at the crack front are linear-elastic quasi-brittle. The fracture strength is statistically distributed and the da/dN curve represents its distribution function. For higher loads and independent from the environment crack tip blunting is also active. The transition seen in the da/dN curve represents the probability that the crack front is blunting. The micromechanical parameter for this mechanism is the striation spacing seen on crack surfaces. The probability functions for 7075-T651 have been determined and will be discussed.

1 INTRODUCTION

Microstructural parameters influencing the fatigue crack growth process like the distance between precipitates, the grain size and specially the grain orientation are randomly distributed. The assumption that each cycle of ΔK_{eff} causes a Δa increment along the full crack front is in general not valid. Especially near threshold the crack is growing irregularly. This has been found in fractographic examinations of crack surfaces, for example in [1]. If the crack propagation is monitored at a fixed location at the crack front, e.g. at the front or the back side of the specimen, then the crack growth rate at constant ΔK_{eff} loading shows a big scatter. This scatter results from the different configurations of the grain structure relative to the moving crack tip.

The crack front can be divided into many segments. Each crack front segment is limited by the grain boundaries of the grain crossing the crack front. For a single segment, the grain orientation and the distance from the crack tip to the back face grain boundary are given. These microstructural parameters define the local material resistance for the crack front segment. Instead of a Δa increment along the full crack front, there is only a certain probability that a crack segment moves one Δa step forward, see Figure 1. Each crack front segment has this probability to move or to stay. However these probabilities along the crack front are not independent of each other. Once a segment is one step further than its neighbors his own probability to move the next step is lowered and the probability of the neighbors to join the segment ahead is increased. This coupling effect of the segments along the crack front is strong. This can be seen in the fact that the crack shape remains smooth when the crack is propagating through the specimen. Khen and Altus have incorporated this coupling effect in their numerical model and they came to the conclusion that the crack length distribution in the long crack regime is independent of the elements interaction probability and is given by the strength distribution only [2]. Due to this strong coupling effect, the

local probabilities are not needed to be formulated along the crack front. The fatigue crack process can be modeled based on an average of these probabilities.

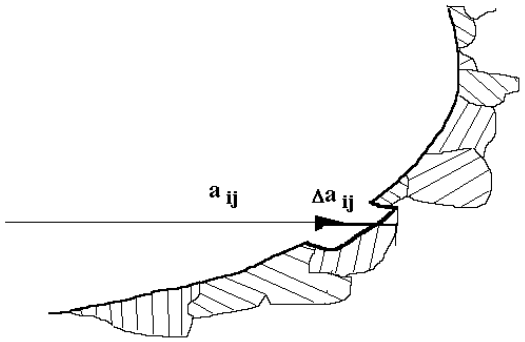

Figure 1: Microstructural configuration at the crack front: The crack front cuts many grains with different orientations and grain length; the crack is moving in one crack front segment a_{ij} by Δa_{ij}.

The micromechanical function of the local resistance against crack advance and the Δa step in a crack front segment should be closely related to the macroscopical loading parameter ΔK_{eff} and the average crack growth rate da/dN. A similar relationship between a microstructural and a macromechanical material behavior has been given by Christ [3] for the stress-strain behavior in metals. This Masing model will be presented first before an analogous model is presented for a crack under cyclic loading.

2 THE NUMERICAL MODEL

1.1 The Original Masing Model

The elastic-plastic deformation behavior of a metal in the one-dimensional stress state can be measured with the stress-strain curve. This is a macroscopical description of the material behavior, which does not distinguish between the many grains in the metal. The macroscopic material behavior is understood to represent the average of many individual microstructural elements. Each element has a very simple behavior and in the case of the Masing model they are connected in parallel. The single element is an ideal elastic perfectly plastic body. At each strain level not every element is in the same situation what yielding is concerned. The weak elements are already yielded because their yielding limit is exceeded. The strong elements are still in the linear elastic regime. The relation between the microscopical material strength distribution function f_p and the macroscopic stress-strain curve is given below:

$$f_p(\varepsilon \cdot E) = -\frac{1}{E^2} \cdot \frac{d^2\sigma}{d\varepsilon^2} \tag{1}$$

In this model the single event is a microscopical plastic deformation and the macroscopic effect is the deviation of the stress-strain curve from the ideal elastic line.

1.2 The Masing Model Adapted to the Threshold for Fatigue Crack Growth

A Masing model can be formulated for the cyclic behavior of a cracked body. The macroscopic material behavior is given by the average crack growth rate da/dN of a constant amplitude loading. On a microscopic scale the material consists of many segments. A row of segments immediately ahead of the crack front is loaded by the singular stress field given by the stress intensity factor range ΔK_{eff}. In the rows behind the crack front the segments have been separated, see Figure 2. Each segment has a quasi-brittle characteristic.

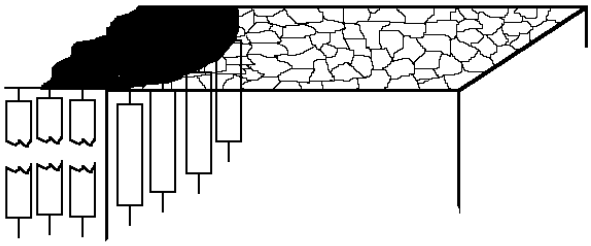

Figure 2: Micromechanical configuration at the crack front and the modeling with single elements

1.2.1 The Fatigue Crack Growth Threshold in Vacuum

In vacuum, the cyclic slip process gives the near threshold crack growth rate. The interaction of the cyclic slip process and the interatomic bond rupture of each single element can be explained as follows: Below a local critical stress, the crack tip reacts purely elastically and no crack growth results. This is the definition of the fatigue crack growth threshold. At the critical shear stress dislocations are free to move in the slip system crossing the crack tip. The shear deformation in their slip plane results in atomic bond rupture at the crack tip. The pair of atoms at the crack tip has lost their contact. The separation of a single element is controlled the local shear deformation in the slip plane. Due to the statistical distribution of the orientation of the grains the local resistance against dislocation motion is statistically distributed. If we define a variable $\Delta K_{j,slip}$ as the quantity for this local resistance of the element j against dislocation motion, crack tip advance can be formulated as follows:

$$\Delta a_{ij} = \begin{cases} \Delta a_{pl} & if \quad \Delta K_{eff,i} \geq \Delta K_{j,slip} \\ 0 & if \quad \Delta K_{eff,i} < \Delta K_{j,slip} \end{cases} \tag{2}$$

The local resistance against dislocation motion is depending on microstructural parameters and therefor $\Delta K_{j,slip}$ is statistically distributed:

$$P_{slip}(\Delta K_{eff}) = \int_0^{\Delta K_{eff}} p_{slip}(\Delta K_{i,slip}) d\Delta K_{i,slip} \tag{3}$$

P_{slip} is the distribution function. It shows the probability that dislocations are free to move on the slip system at position j. Because this distribution function is independent of the position in the material, it is also valid along the crack propagation direction [4]. The crack growth rate is the average of all the elementary crack growth steps:

$$da/dN\left(\Delta K_{eff}\right) = \int_0^{\Delta K_{eff}} \Delta a_{pl} \cdot p_{slip}(\Delta K_{i,slip}) \cdot d\Delta K_{i,slip} + \int_{\Delta K_{eff}}^{\infty} 0 \cdot p_{slip}(\Delta K_{i,slip}) \cdot d\Delta K_{i,slip}$$

$$= \Delta a_{pl} \cdot \int_0^{\Delta K_{eff}} p_{slip}(\Delta K_{i,slip}) \cdot d\Delta K_{i,slip} = \Delta a_{pl} \cdot P_{slip}\left(\Delta K_{eff}\right) \tag{4}$$

Where the probability density function is assumed to be of Weibull type [5]:

$$P_{slip}\left(\Delta K_{eff}\right) = 1 - e^{\left[1-\frac{\Delta K_{eff}}{\Delta K_{th,slip}}\right]^{1/\alpha_{slip}}} \tag{5}$$

1.2.3 The Fatigue Crack Growth Threshold in Corrosive Environments

In the case of the fatigue crack growth behavior in corrosive environments a similar approach can be used as for the case in vacuum. Instead of the cyclic slip process, the oxide film fracture mechanism is controlling the near threshold regime. The local critical stress intensity factor range is given by the ultimate strength of the oxide film $\Delta K_{j,ox}$.

$$\Delta a_{ij} = \begin{cases} d_{ox} & if \quad \Delta K_{eff,i} \geq \Delta K_{j,ox} \\ 0 & if \quad \Delta K_{eff,i} < \Delta K_{j,ox} \end{cases} \tag{6}$$

Similar to the derivation for local cyclic slip mechanism, the crack growth rate can be found as:

$$da/dN\left(\Delta K_{eff}\right) = d_{ox} \cdot \int_0^{\Delta K_{eff}} p_{ox}(\Delta K_{i,ox}) \cdot d\Delta K_{i,ox} = d_{ox} \cdot P_{ox}\left(\Delta K_{eff}\right) \tag{7}$$

And again the probability density function is assumed to be of Weibull type:

$$P_{ox}\left(\Delta K_{eff}\right) = 1 - e^{\left[1-\frac{\Delta K_{eff}}{\Delta K_{th,ox}}\right]^{1/\alpha_{ox}}} \tag{8}$$

1.2.4 The Fatigue Crack Growth Threshold of the Blunting Mechanism

The transition from near threshold regime to the Paris regime is a threshold for the blunting process. The da/dN versus ΔK_{eff} relation near this transition represents the distribution function of the local resistance against blunting $\Delta K_{i,bl}$. In the case of vacuum this results in:

$$da/dN\left(\Delta K_{eff}\right) - \Delta a_{pl} = \Delta a_{str} \cdot \int_0^{\Delta K_{eff}} p_{bl}(\Delta K_{i,bl}) \cdot d\Delta K_{i,bl} = \Delta a_{str} \cdot P_{bl}\left(\Delta K_{eff}\right) \quad (9)$$

Where

$$P_{bl}\left(\Delta K_{eff}\right) = 1 - e^{-\left[1-\frac{\Delta K_{eff}}{\Delta K_{th,bl}}\right]^{1/\alpha_{bl}}} \quad (10)$$

3 DETERMINATION OF THE MODEL PARAMETERS

The model parameters are found by adjusting the da/dN functions given in eqns (7) and (9) to the da/dN data found in constant amplitude tests.

The model parameters for 7075-T651 in air, nitrogen and vacuum are given in Table 1. The data in nitrogen allow us to choose clearly the model parameters. Figure 3 indicates that reasonable values for the parameters could be found, such that the model fits well with the measured crack growth rates.

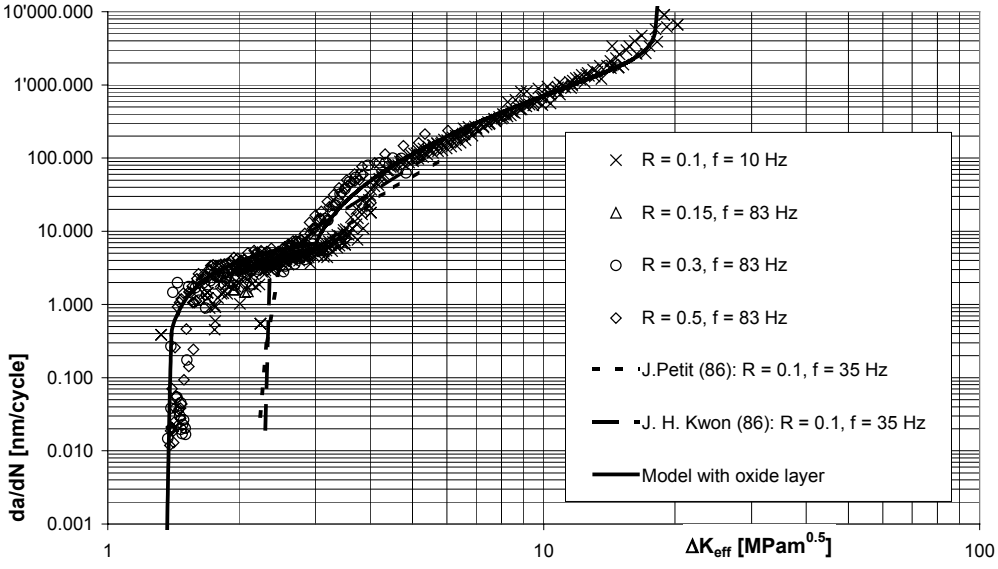

Figure 3: Fitted Probability Functions and Crack Growth Rate Data for 7075-T651 in air

Table 1 Parameters of the Probability Functions

Parameter	7075-T651 in Air	7075-T651 in Nitrogen	7075-T651 in Vacuum	Unit
$\Delta K_{th,slip}$	-	-	2.65	MPa$\sqrt{m}$
$\Delta K_{th,ox}$	1.35	1.40	-	MPa$\sqrt{m}$
$\Delta K_{th,bl}$	3.00	4.10	4.50	MPa$\sqrt{m}$
α_{slip}		-	2.00	-
α_{ox}	0.75	0.10	-	-
α_{bl}	0.75	1.50	1.75	-

4 DISCUSSION

The irregular crack growth behavior near threshold has been model by a numerical approach. The average crack growth rate is interpreted as the product of a micromechanical length, e.g. the distance between slip planes or the oxide film thickness and a probability. Crack growth rate data has been used to find appropriate parameters for the probability functions. The transition seen at higher load levels is also seen as a threshold. The mechanism initiated in this regime is crack tip blunting with the fatigue striation spacing as the relevant micromechanical length. The proposed model is able to explain the deviations found in the da/dN data from classical theoretical models for crack growth, such as the Paris equation or a Forman fit.

5 ACKNOWLEDGEMENTS

The financial support for this work from the Science and Technology Group and the Aircraft Group at the Swiss Procurement Agency is gratefully acknowledged.

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