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Calculating the fatigue life of smooth specimens of two-phase titanium alloys subject to symmetric uniaxial cyclic load of constant amplitude

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ABSTRACT

A system of microstructure-dependent fatigue failure models that allows calculating the number of stress cycles to failure of smooth specimens in the absence of data on the fatigue resistance of the material is proposed. Fatigue life is represented as the sum of the number of stress cycles to the initiation of a microstructurally short crack with a depth equal to one grain size and the number of stress cycles to the instant the growing physically small and long crack reaches the depth taken as a fatigue failure criterion. To fill the models, it is necessary to test standard specimens under monotonic short-term tension to determine the elastic modulus, Poisson's ratio, and the proportional limit and to analyze the microstructure and texture of the material to determine the Taylor factor, the magnitude of the Burgers vector, and the distance between neighboring parallel slip planes in the lattice, depending on which slip system is activated relative to the tension direction. The models are applicable to metals and alloys in which a fatigue crack is initiated along a persistent slip band in a surface grain under high-cycle loading. The models are validated against fatigue test data for flat specimens of different two-phase titanium alloys (VT3-1, IMI 834, Ti–6Al–4V, VT6, VT16, VT22, VT23, LCB) with different types of microstructure (bimodal, globular, fine-grained β-transformed) subject to symmetric in-plane bending. The plotted *S–N* curves are in a satisfactory agreement with experimental data.

1. Introduction

The present generally accepted idea about high-cycle fatigue is that the fatigue process occurs in two stages: (i) dispersed fatigue damage (many microscopic centers of damage – local plastic deformations) and (ii) local fatigue damage (main fatigue crack growing to final failure). In this connection, stage (i) for certain metals and alloys, including titanium alloys, may be regarded as accumulation of dislocations in slip planes of surface grains favorably oriented relative to the loading direction [1–7]. As a result, this process gives rise to microstructurally short cracks (MSC) that are no larger than the grains and are aligned along persistent slip bands (PSB). At stage (ii), one of these initiated plane MSCs that is favorably oriented relative to neighboring grains continues to grow as a mixed-mode physically short crack (PSC), several grain sizes deep, at some transitional stage, turning into a main mode I long crack (LC), which leads to final fatigue failure [7–14].

The important role of the microstructure in the nucleation and growth of fatigue cracks has been generally recognized, and there are many studies on the subject [2–6,10–13,15]. It is clear that the fatigue models that describe the behavior of the material at stages (i) and (ii) must be based on different approaches. The major shortcoming and, hence, limitation of the available fatigue crack initiation life models is their incapability of predicting the crack length at the instant of crack initiation. Thus, this length must be prescribed as an initial condition for crack growth models.

The present paper reports the results of studies intended to improve the crack initiation and growth models that explicitly include crack length and microstructure parameters and that could be used to calculate fatigue life (*S–N* curves) of smooth specimens at normal temperature and under unidirectional cyclic loading. The calculated *S–N* curves will be validated against the experimental data on the fatigue resistance of two-phase titanium alloys [16–18].

2. Fatigue crack initiation life model

2.1. Available microstructure-dependent models

Fatigue crack initiation life models that explicitly include microstructure parameters have been developed in the last three

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Nomenclature

a, c, h	hexagonal close-packed crystal lattice parameters
$\vec{b}$	Burgers vector
b	Burgers vector module
c	surface crack half-length
d	grain size
d_o	average thickness of grain boundary α-layer
d_1	average diameter of the primary α-phase particles and thickness of α-lamellas
d_2	reduced size of the grain boundary α-layer
D_β	average diameter of β-grains
dl/dN	fatigue crack growth rate
E	elastic modulus
G	shear modulus
k	critical resolved shear stress
K	stress intensity factor
K_I	maximum stress intensity factor of the mode I
K_1	local stress intensity factor of the mode I
K_2	local stress intensity factor of the mode II
K_{eq}	equivalent stress intensity factor
K_T	stress intensity factor for the physically small-to-long crack transition
K_{th}	fatigue threshold stress intensity factor for a long crack
$K_{th,d}$	fatigue threshold stress intensity factor for a microstructurally short crack
$K_{th,l}$	fatigue threshold stress intensity factor for a physically small crack
l	crack depth
l_i	intermediate physically small crack depth
l_i'	physically small crack depth at the transition point to long crack
l_s	intermediate physically small crack depth at $\sigma_a = \sigma_{-1}$
l_t	final long crack depth (fatigue failure criterion)
M	Taylor factor
m_s	Schmid factor
N	number of load cycles
N_i	number of load cycles to crack initiation
N_{FCG}	number of load cycles during the crack propagation
N_{total}	number of load cycles to failure
r_{pc}	cyclic plastic zone size
Y, Y_1, Y_2	geometry factors (stress intensity factor corrections)
W_s	specific fracture energy per unit area along a slip band
δ	slip band width, crack tip opening displacement
γ	slip direction-load direction angle
φ	slip plane normal-load direction angle
λ	universal constant
ν	Poisson's ratio
$\theta, \theta_o, \theta_1$	orientation angles
ρ	crack tip radius
σ	normal stress
σ_a	cycle stress amplitude
σ_f	internal friction stress in the crystal lattice
σ_{max}	maximum local stress
σ_p	proportionality limit
σ_{-1}	fatigue limit at a symmetric load cycle
$\Delta\sigma$	tensile stress range
$\Delta\tau$	shear stress range
ζ	misorientation parameter
CRSS	critical resolved shear stress
CTOD	crack tip opening displacement
FGBT	fine-grained β-transformed
HCP	hexagonal close-packed
LC	long crack
LEFM	linear elastic fracture mechanics
MSC	microstructurally short crack
PSB	persistent slip band
PSC	physically small crack
SIF	stress intensity factor

decades. Tanaka and Mura [4] propose dislocation models for predicting crack initiation in slip bands, inclusions, and notches. They believe that a fatigue crack results from the accumulation of dislocation dipoles during cyclic deformation. The initiation of a crack along a slip band is modeled as follows [4]:

$$(\Delta\tau - 2k) \cdot N_i^{1/2} = \left[\frac{8GW_s}{\pi d}\right]^{1/2}, \tag{1}$$

where $\Delta\tau$ is the shear stress range, k is the critical resolved shear stress (CRSS), N_i is the number of load cycles to crack initiation, G is the shear modulus, d is the grain size, W_s is the specific fracture energy per unit area along a slip band. This model was improved by Chan and presented in [5,6]. The initiation of a fatigue crack along PSB in polycrystalline materials is modeled in terms of applied tensile stress range $\Delta\sigma$ as follows [5]:

$$(\Delta\sigma - 2Mk) \cdot N_i^{\alpha} = \left[\frac{8M^2G^2}{\lambda\pi(1-\nu)}\right]^{1/2} \left(\frac{\delta}{d}\right) \left(\frac{c}{d}\right)^{1/2}, \tag{2}$$

where M is the Taylor factor ($M = 1/m_s$, m_s is the Schmid factor), ν is Poisson's ratio, $\lambda = 0.005$ is a universal constant, δ is the slip band width, c is the crack depth or the surface crack half-length.

To use model (2) to calculate the fatigue crack initiation life N_i, it is necessary to determine the material constants G, ν and the grain size d. The term $2Mk$ is considered, by the author of model (2), to be a fatigue limit below which no crack is initiated. These parameters and the fatigue limit should be derived from experimental data or found in the literature. The Taylor factor M is generally dependent on the texture, but is equal to 2 for the most favorably oriented grain; i.e., the angle γ between the normal $\vec{n}$ to the slip plane and the loading direction and the angle φ between the slip direction $\vec{b}$ and the loading direction (Fig. 1) are equal to $\pi/4$, i.e., $m_s = \cos\gamma \cdot \cos\varphi = 1/2$. The exponent α generally falls between 0 and 1 and depends on the stacking-fault energy and the amount of slip irreversibility. The value $\alpha = 0.5$ provides the best agreement with experimental data for many metals and alloys, including titanium alloys [5]. Model (2) includes two more unknown quantities: crack depth c and slip band width δ or the crack width at the instant of crack initiation [5]. The parameter δ

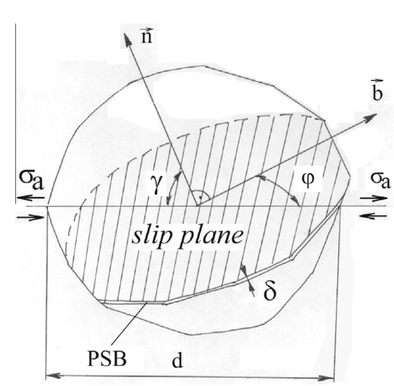

Fig. 1. Orientation of the slip plane in a grain relative to the loading direction.