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A new approach to estimating the fatigue notch factor of Ti-6Al-4V components

Gbadebo Owolabi *, Oluwamayowa Okeyoyin, Adewale Olasumboye, Horace Whitworth

Department of Mechanical Engineering, Howard University, Washington, DC 20059, USA

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ABSTRACT

Titanium alloy is used for airframe components and compressor blades application because of its high strength and fracture toughness at low temperatures and high strength and creep resistance at elevated temperature. This paper extends a recently developed probabilistic mesomechanics based model to notched titanium alloy components using simulation strategies that capture both the essence of notch root stress gradient and the complexity of realistic microstructures. The notch size effects and notch root and inelastic behavior are combined with probability distributions of microscale stress and small crack initiation to inform minimum life design methods. A new approach which can be applied using crystal plasticity finite element or closed form solution is also proposed as a more robust method for determining the fatigue notch factor than the existing classical methods. The fatigue notch factors predicted using the new probabilistic mesomechanics based model are in good agreements with experimental for notched titanium alloy specimens subjected to cyclic loads with various stress ratios.
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1. Introduction

Titanium alloy Ti-6Al-4V offers a range of properties such as high strength and fracture toughness at low temperatures to high strength and creep resistance at elevated temperature. Thus, it is widely used in engineering applications from airframe components and fans to compressor blades of jet engines. Ingestion of debris into the engine of aircraft during takeoff and landing causes nicks and dents to form on the leading and trailing edge of turbine blades (Fig. 1). These dents and nicks can be treated as small notches with a notch root radius and notch depth and thus serve as stress raiser and favorable zones for crack initiation therefore reducing the fatigue strength of the material [1–3]. The fatigue strength reduction factor otherwise known as the fatigue notch factor, k_f, is used in the estimation of fatigue life and strength of notched structural components. Several models based on different assumptions have been developed for k_f in the past. The Neuber [4], Kuhn and Hardraht [5], Peterson [6], Heywood [7,8], Buch [9,10], and Siebel and Stieler [11] models are all based on average stress assumptions. One of the drawbacks of these models is that they do not incorporate explicit sensitivity to the combined effects of microstructure and strength of the notch root stress field gradient. Recent approaches have been developed to incorporate the stress gradient field at notches [12,13], but are deterministic and do not address the role of microstructure explicitly [14]. It is therefore very difficult to link the k_f obtained using these methods to the realistic microstructure of the material such as grain size, grain orientation, multi-phases, and the notch root geometry.

Owolabi et al. [14,15] recently developed probabilistic models based on the weakest link theory and extreme-value statistics where elements of crystal plasticity were combined with new probabilistic methods for notch sensitivity based on computed slip at the notch root within a well-defined fatigue damage process zone for homogenous oxygen free high conductivity copper. The purpose of this paper is to extend this model to heterogeneous dual-phase aero-engine materials such as the titanium alloy. The framework will incorporate information regarding not only the peak stress but also the stress gradient relative to microstructure length scale. This approach can reduce the amount of testing required to make design decisions on material or component reliability by systematically estimating the scatter of fatigue life associated with microstructure variations through the use of simulations.

2. Material and crystal plasticity models

Titanium is widely known for its good resistance to corrosion and high strength to weight ratio. When alloyed with other metals and heat treated, it can achieve a wide range of attractive proper-

* Corresponding author. Tel.: +1 202 806 6594; fax: +1 202 483 1396.
E-mail address: gbadebo.owolabi@howard.edu (G. Owolabi).

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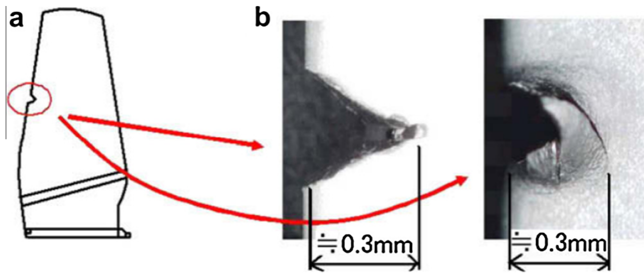

Fig. 1. Image of foreign object damage: (a) fan blade schematics and (b) FOD damage example on edges of airfoil [2].

ties both at low and high temperatures. Titanium alloys exhibit high strength and creep resistance at high temperatures while at low temperatures, they exhibit high strength and fracture toughness. At room temperature, an unalloyed titanium has an hcp crystal structure but with increase in temperature to the neighborhood of 882 °C, it undergoes a change in phase from the hcp α phase to a bcc β phase. It is possible to produce a completely α-phase, β-phase, or a mixture of α and β phases by varying the alloying elements. Different microstructures can be realized for the same alloy by varying the processing parameters and heat treatment. Titanium alloy is dual phase and it has the inherent problem of varying individual phase properties due to different amount of alloying elements. The hcp structure of titanium, unlike for fcc structured materials, has several planes which are favorable for occurrence of slip or twinning. In most hcp materials, basal (0001) and prismatic $\{10\bar{1}0\}$ have been identified as the primary slip planes with a closed packed direction $\left\langle 11\bar{2}0\right\rangle$ for the slip vector. Titanium has been identified to exhibit flow stress versus temperature anomaly as well as orientation dependent yielding according to the work of Naka et al. [16].

The crystal plasticity model used in this work followed the existing work of Mayeur and McDowell [17], Zhang et al. [18] and Bridier et al. [19] for Ti-6Al-4V. The duplex Ti-6Al-4V is a dual-phase alloy consisting of an hcp structured matrix primary α-phase and secondary lamellar α + β domains with alternating layers of secondary α laths and bcc structured residual β laths. The crystal plasticity model accounts for the distinct three-dimensional slip geometry for each phase, anisotropic and length scale dependent slip system strengths, the non-planar dislocation core structure of prismatic screw dislocations in the primary α-phase, and crystallographic texture [18]. The relationship between the slip system shearing rate and the resolved shear stress of the α slip system is described by the power law flow rule given as:

$$\dot{\gamma}^\alpha = \dot{\gamma}_0 \left\langle \frac{|\tau^\alpha - \chi^\alpha| - \kappa^\alpha}{D^\alpha} \right\rangle^m \mathrm{sgn}(\tau^\alpha - \chi^\alpha) \qquad (1)$$

where $\dot{\gamma}_o$ is the reference shearing rate, m is the flow exponent which controls the rate sensitivity of flow, τ^α is the resolved shear stress, χ^α is the back stress, κ^α is the length scale-dependent threshold stress and D^α is the drag stress. The drag stress is taken as a non-evolving constant, i.e. $\dot{D}^\alpha = 0$, while the back stress evolves according to an Armstrong–Frederick direct hardening/dynamic recovery type of equation [19], i.e.

$$\dot{\chi}^\alpha = h\dot{\gamma}^\alpha - h_D \chi^\alpha |\dot{\gamma}^\alpha| \qquad (2)$$

with $\chi^\alpha(0) = 0$. The threshold stress is expressed as [19]:

$$\kappa^\alpha = \frac{\kappa_y}{\sqrt{d^\alpha}} + \kappa_s^\alpha \qquad (3)$$

In Eq. (3), the first term uses a Hall–Petch-type formulation with d^α as the microstructural dimension of the free slip length and κ_y as the Hall–Petch constant. The second term corresponds to the softening parameter that addresses the breakdown of short range order due to the dislocation glide commonly observed in a-Ti-6Al-4V at room temperature where planar slip of screw dislocations promotes a nearly elastic-perfectly plastic behavior [19]. The threshold stress evolution law based on softening of all slip systems is of the form [19]:

$$\dot{\kappa}_s^\alpha = -\mu \kappa_s^\alpha |\dot{\gamma}^\alpha| \qquad (4)$$

where μ is a constant.

It is noted that the cyclic behavior of titanium dual-phase titanium alloy depends significantly on the activity of twinning and the interaction of dislocations with twins [20]. The deformation microstructure at room temperature depends, however, on the applied strain amplitude. Thus, for the purpose of this study only slip is assumed to be the main mechanism of plastic deformation. This assumption is reasonable since under low strain amplitudes as in the case of high cycle fatigue considered here, the influence of twinning is minimal; thus, the primary deformation mechanism in Ti-6Al-4V (like all α/β Ti-Al alloys) is the prismatic glide of $\langle a\rangle$-type screw dislocations [17,20]. Moreover, for Ti-6Al-4V alloy, William et al. [21] have shown that the occurrence of twinning decreases with increasing aluminium content and in titanium alloy with 6% or higher atomic wt.% aluminium, twinning is hardly observed [20]. Thus, with the role of twinning being negligible in many αβ alloys, the importance of **c** + **a** slip is enhanced since it is the sole mechanism responsible for accommodating deformation along the *c*-axis [20].

It is very difficult to explicitly model the α + β colonies containing the secondary α and β phases arranged in lamellar structure since the thicknesses of the laths are much smaller structures when compared to the microstructural statistical volume element of the titanium alloy using a finite element mesh that must include a large number of grains [19]. To model the α + β colonies, we followed the approach detailed in Refs. [17–20] where the slip geometry of lamellar colonies is obtained by homogenizing the lamellar structure systems in an equivalent grain-scale representation of the α + β colonies based on the burgers orientation relation (BOR) given by $(0001)_\alpha // \{101\}_\beta$ and $(11\bar{2}0)_\alpha // \{111\}_\beta$ as shown in Fig 2. There are 24 possible slip systems in the lamellar region: $3\langle 11\bar{2}0\rangle(0001)$ basal, $3\langle 11\bar{2}0\rangle\{10\bar{1}0\}$ prismatic, $6\langle 11\bar{2}0\rangle\{10\bar{1}1\}$ first order pyramidal and $12\langle 111\rangle\{110\}$ bcc slip systems which are transformed into the hexagonal coordinate system according to the BOR [18]. The readers are referred to Refs. [17–19] for more details on this modeling approach.

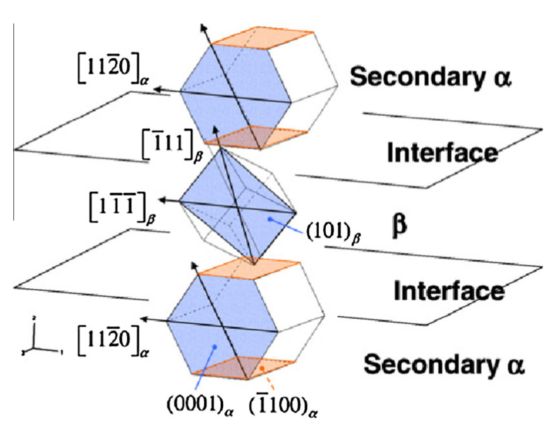

Fig. 2. Lamellar orientation relationships [18].

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