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Phase morphology, variants and crystallography of alloy microstructures in cold dwell fatigue

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This paper examines microscale crystal slip accumulation, cold creep, and stress redistribution (load shedding) related to dwell fatigue in a range of α–β Ti alloy microstructures. The role of basal slip and prism slip is evaluated in load shedding in a rogue grain combination. The results enrich the Stroh dislocation pile up interpretation of dwell by accounting for the anisotropic rate dependence of differing slip systems together with morphology.

Microstructural morphology has been found to play an essential role in cold creep and load shedding in dwell fatigue. Basketweave structures with multiple α variants have been shown to give the lowest load shedding for which the mechanistic explanation is that the β lath structures provide multiple, small-scale α variants which inhibit creep and hence stress relaxation, thus producing more uniform, diffuse stress distributions across the microstructure through *microscale kinematic confinement,* imposed by multi (α)-to-single (β) BOR relations (i.e. multiple α variants sharing the same parent β grain). The critical consequence of this is that alloys typically having multi-variant basketweave structure (e.g. Ti-6246), remain free of dwell fatigue debit whereas those alloys associated with globular colony structures (e.g. Ti-6242) suffer significant dwell debit. This understanding is important in microstructural design of titanium alloys for resisting cold dwell fatigue.

1. Introduction

Component failure due to cold dwell fatigue was first recognized from in-service behaviour in RR RB211 engine Ti-alloy discs on Lockheed Tristar aircraft in the 1970s [42]. In this problem, the defining characteristics is that the inclusion of a dwell period in the stress loading history is found to reduce the fatigue life substantially (sometimes orders of magnitude), and has been argued to be through cyclic accumulation of plastic strain during cold creep [5] and Lefranc et al. [28,29] both in air as well as in high vacuum. The mechanisms of fatigue facet nucleation in titanium alloys have been studied by several authors including Dunne and Rugg [14]. In the latter, the load shedding within a rogue grain (soft-hard) combination was thought to play a significant role in facet fracture near basal planes [47,56], and which potentially leads to extreme values in fatigue indicator parameters [40,41].

The Stroh model [50] for the stresses developing at the termination of an active slip band at a grain boundary has been used to explain fatigue facet formation in titanium alloys by many authors [5,17]. However, the Stroh model was based on 2D isotropic elastic plane theory [48], and does not consider either crystal slip nor its strain rate sensitivity (SRS) giving rise to local creep even at low temperature. Multiple slip systems, but predominantly basal and prism systems, may be activated by applied loading in α-β titanium alloys [8], and the intrinsic anisotropy of multiple slip systems operating may be associated with the pile up of dislocations. Pilchak et al. [37,38] addressed non-dwell facet growth in Ti–6Al–4V colonies and found facets growing along basal planes, although deviations as high as 35° were also measured. In the context of rogue (hard-soft) grain combinations, researchers have often assumed prismatic slip in the soft grain adjacent to the hard grain [34], since it has been believed to be more readily activated in titanium alloys. In contrast, Evans and Bache [17] proposed a scenario where the dislocation pile-up occurs on a basal slip plane in a soft grain which potentially leads to facet nucleation, and this has also been discussed by Sinha et al. [48]. In addition, faceting on the adjacent basal plane of a hard grain has been identified as the most critical damage mode in leading to fracture in a commercial α/β-forged Ti–6Al–4V alloy [9]. An explanation was proposed recognising the resolved shear stress, reflecting the local Schmid factor and the normal stress in relation to the elastic anisotropy of the α-phase. It has also been found that the elastic and plastic anisotropies are important in material rate dependence and in localization of plastic slip and

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microcrack nucleation [12,36].

An anisotropic rate dependence in differing slip systems and phases has recently been reported for titanium alloys [59,60] where basal slip has been found to be intrinsically more rate sensitive at 20 °C than for prism slip. Pilchak [39] pointed out that in order to characterize the dwell fatigue crack initiation, the strain rate sensitivity and strain accumulation during the dwell period have to be recognised. The strain rate sensitivity is itself strongly dependent on microstructure which in turn primarily results from the bulk deformation forming and heat treatment processes utilised for the material manufacture. Microtextured regions, or macrozones are also a consequence for forming history. Woodfield et al. [56] have pointed out that material microstructures in the Ti alloys are critically related to the component dwell fatigue life.

Processing route in the Ti alloys potentially leads to a substantial range of microstructures, including primary alpha, colony, bi-modal and Widmanstätten structures, resulted in very different uniaxial yield and ultimate strengths, tensile ductility [49] and crack development [32]. Microstructures consisting of large regions of aligned, crystallographically oriented alpha plates, often referred to as macrozones [20] have been argued to be most susceptible to dwell fatigue [19,54,56]. Britton et al. [10] noted that geometrically necessary dislocation density was found to be twice as high in the macrozone region than outside of it. However, macrozones have been suggested to contribute more significantly to crack coalescence and growth rather than to crack nucleation [27,30].

With differing cooling rate, dual phase (α-β) titanium alloys may have remarkably different morphologies. For example, a cooling rate of 1 °C/min may lead to colony structures but a much faster rate of 8000 °C/min produces a Widmanstätten basketweave structure [31]. Air cooling of near-α Ti-6242 produced by α/β processing may generate equiaxed structures [26]. Interestingly, basketweave microstructures, containing multiple colonies of α and β laths within prior β grains, were more creep resistant than the bimodal microstructure in Ti-6242Si [7]. The samples with basketweave structure have been found not to be dwell sensitive in β processed Timetal 685 [16], or in Ti-6242 alloys [51]. Miller et al. [33] found creep strains to be smaller in basketweave structures compared to colony structures in Ti-6Al-2Nb-1Ta-0.8Mo alloy. They argued that larger colony sizes led to higher creep strains and that smaller colony size provided shorter distances to accommodate slip, and that the basketweave structure exhibited limited slip compatibility. In addition, the thickness of α platelets has been reported to affect creep behaviour [46]. Recent pillar tests also show that thicker β laths produce more impedance to slip transfer through α-β-α phase boundaries than thinner laths [59] and the basketweave structure is less rate sensitive in α-β titanium [57] than that for colonies.

In addition, alloy Ti-6246 often used for compressor discs in engines has been found to have less dwell sensitivity as opposed to Ti-6242 which has a very strong dwell debit [43], although it is noted that this may relate more to the resulting microstructures in each alloy as opposed to their chemistry differences. Hence the recent studies showing that Ti-6246 exhibits a significant time-dependent plastic strain accumulation, which is comparable to Ti-6242 during the stress holds [22], are interesting. Very recent work by Zhang and Dunne [57] finds the strain rate sensitivity to be lower in the basketweave structures, specifically in Ti-6246 with a greater number of α variants, than the usual colony structures in Ti-6242. Note that basketweave microstructures could have up to twelve α variants sharing a single β grain (multi (α)-to-one (β) BOR relation). In contrast, the colony structure has sandwiched α-β phase laths with each α-β interface having the same one (α)-to-one (β) BOR relation. These findings suggest an explanation for the long standing problem in dwell debit difference between alloys Ti-6242 and Ti-6246. The former is typically near-α (β volume fraction less than 15%), and most often has colony and lamella structures. Ti-6242 may also take a basketweave form which has been reported to have greater creep resistance than colony structures [51]. Ti-6246, however, an α-β alloy having up to 45% β [3,24], is known to be dwell insensitive. A potentially important factor is that the greater β phase volume fraction is accompanied by more α variants thus likely to provide stronger *microscale kinematic confinement* effects.

Hence it is suggested that the dwell sensitivity in commercial Ti alloys depends on multiple factors but that particularly, α-β microstructure (including morphology as well as crystallography) plays a critical role [57]. It is argued herein that the alloy chemistry plays a lesser part, so that previous hypotheses that Ti-6242 is dwell sensitive whereas alloy Ti-6246 is not are over-general and do not recognise the importance of processing giving rise to microstructural differences. However, when the microstructural differences from processing are reflected in the material behaviour, then strong differences in dwell behaviour are indeed observed [4,6,43,57,59,61]. Hence, the effectiveness of differing microstructures needs to be assessed systematically in cold creep behaviour both at the grain or structural unit-level and at the transgranular or polycrystalline-level so as to provide useful insights into cold dwell fatigue and the optimisation of dwell-resistant alloys.

This paper presents a systematic study of the slip accumulation, constraint and stress redistribution (load shedding) in key Ti microstructures from pure α systems through to α–β colonies, Widmanstätten and basketweave structures. Basketweave microstructures are here differentiated from Widmanstatten as those containing multiple α variants compared to the single variant in the latter structures, whilst all variants satisfy the separate Burger Orientation Relationship (BOR). In addition, in the α–β microstructures, the β thickness with respect to the α laths is also investigated in respect of cold creep and stress redistribution. The methodology is to employ the dislocation based crystal plasticity modelling method in which basic phase property data has been established from direct micro-pillar testing to give intrinsic slip strengths and phase strain rate sensitivities.

2. Framework of crystal plasticity modelling

In this study, we adopt a rate dependent crystal plasticity with updated lattice rotation [15,59]. The strain hardening is included through the evolution of geometrically necessary (GND) and statistically stored (SSD) dislocation density. This allows for the proper representation of the strain gradient effect introduced by grain/phase boundaries of multiphase polycrystal materials.

2.1. Kinetics

The multiplicative elastic-plastic decomposition of the deformation gradient is given by

$$\mathbf{F} = \mathbf{F}^{e}\mathbf{F}^{p} \tag{1}$$

The plastic deformation gradient develops as

$$\dot{\mathbf{F}}^{p} = \mathbf{L}^{p}\mathbf{F}^{p}. \tag{2}$$

The plastic velocity gradient $\mathbf{L}^p$ is that associated with plastic flow through a fixed lattice and is given by

$$\mathbf{L}^{p} = \sum_{i} \dot{\gamma}^{i}\mathbf{s}^{i} \otimes \mathbf{n}^{i} \tag{3}$$

where $\mathbf{s}^i$ and $\mathbf{n}^i$ are updated slip directions and plane normals. The slip rate $\dot{\gamma}^i$ is given by

$$\dot{\gamma}^{i} = \rho_m \nu b_i^2 \exp\left(-\frac{\Delta F^i}{kT}\right)\sinh\left[\frac{(\tau^i - \tau_c^i)\Delta V^i}{kT}\right] \tag{4}$$

where i is the active slip system (i.e. in the α HCP or β BCC phase). ρ_m is the density of mobile dislocations, ν the frequency of attempts of dislocations to jump obstacle energy barriers, b the Burger's vector magnitude, k Boltzman's constant, T the temperature, τ^i the resolved shear