



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

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ABSTRACT

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. Micro-textured 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 Widmanstätten 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

$$\mathbf{F} = \mathbf{F}^e \mathbf{F}^p \quad (1)$$

The plastic deformation gradient develops as

$$\dot{\mathbf{F}}^p = \mathbf{L}^p \mathbf{F}^p, \quad (2)$$

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

$$\mathbf{L}^p = \sum_i \dot{\gamma}^i \mathbf{s}^i \otimes \mathbf{n}^i \quad (3)$$

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

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

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

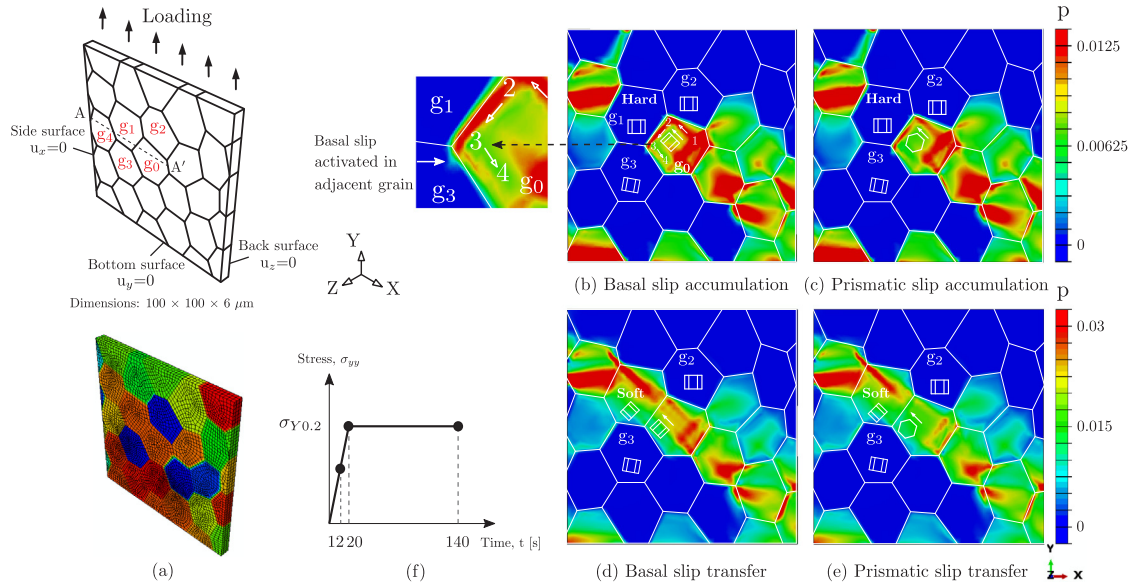


Fig. 1. Accumulated slip due to activated basal and prismatic systems in grain g_0 : (a) model dimensions, mesh, boundary conditions and loading; (b) and (c) are basal and prismatic slip accumulation at the hard grain, respectively, with the insert highlighting the activation of basal slip in the adjacent hard grain g_3 ; (d) and (e) are the basal and prismatic slip development across well-orientated soft grains, respectively. All unmarked grains are assigned orientations randomly which remain unchanged. These results are taken at the end of a two-minute ($t = 140$ s) stress dwell. (f) shows the loading history applied which is the 0.2% proof stress of alloy Ti-6242.

stress for the activated slip system, and τ_c^i is the corresponding critical resolved shear stress (CRSS). The slip rule in Eq. (4) describes the thermally activated escape of pinned dislocations. The quantity ΔF is the activation energy, and ΔV the corresponding activation volume for governing the overall rate sensitive deformation.

In the crystal modelling, the GNDs and SSDs contribute to the strengthening of the effective critical resolved shear stress and consequently the strain hardening [18,53], as follows

$$\tau_c = \tau_{c0} + Gb\sqrt{\rho_G + \rho_s} \quad (5)$$

The plastic spin term is given by the anti-symmetric part of the plastic velocity gradient

$$\mathbf{W}^p = \sum_i \text{asym}(\gamma^i \mathbf{s}^i \otimes \mathbf{n}^i) \quad (6)$$

Therefore the lattice spin is given by

$$\mathbf{W}^e = \mathbf{W} - \mathbf{W}^p \quad (7)$$

where the continuum spin $\mathbf{W}$, is given by the antisymmetric part of the total velocity gradient, $\mathbf{L}$

$$\mathbf{W} = \text{asym}(\mathbf{L}) \quad (8)$$

where $\mathbf{L} = \dot{\mathbf{F}}\mathbf{F}^{-1}$, and $\mathbf{F}$ is the current deformation gradient.

The crystal orientation may then be updated by the lattice spin as

$$\dot{\mathbf{g}} = \mathbf{W}^e \mathbf{g} \quad (9)$$

2.2. The development of GND and SSD densities

The evolution of SSDs is taken simply to be

$$\dot{\rho}_s = \gamma' \dot{p} \quad (10)$$

where γ' is the coefficient for SSD hardening and the rate of accumulated effective plastic strain is $\dot{p} = \sqrt{\frac{2}{3} \mathbf{D}^p : \mathbf{D}^p}$, where $\mathbf{D}^p$ is the rate of crystal plastic deformation [59].

The development of GND density is obtained from the gradient of the plastic part of the deformation gradient based on Nye's dislocation tensor [11,13,35].

$$\sum_{i=1}^N \mathbf{b}_G^i \otimes \rho_G^i = \text{curl}(\mathbf{F}^p) \quad (11)$$

3. Role of morphology in slip accumulation and load shedding in Ti alloys

Rate dependent cold creep leads to stress redistribution from 'soft' grains orientated favourably for slip on to hard grains not well orientated. Microscale stress redistribution takes place as a result of slip with differing rate sensitivity responding to the local stress state. It is therefore appropriate to examine in more detail the activation of slip in the soft grains and to establish the local stress redistribution, or load shedding, which takes place, and the role of more rate-sensitive local basal slip with respect to prismatic slip. This has been a long-standing question of relevance to understanding cold dwell fatigue.

3.1. Basal versus prismatic slip system activation adjacent to hard grain

Soft-hard α grain interactions are governed by dislocation pile-up at the grain boundary which provides local stress build up and hardening. The Stroh [50] pile-up model has often been invoked to explain the stress concentration established at a soft-hard grain boundary by the slip taking place in the soft grain developing a pile-up at the soft-hard grain boundary [17]. However, this model neglected the material rate dependence leading to creep in Ti alloys and hence local grain-level stress redistribution, or load shedding, and in addition did not address the local intrinsic anisotropy in slip system rate sensitivity. Researchers have often assumed the dislocation pile-ups occurred on prismatic slip systems [34]. Recent experiments and numerical studies have shown distinctly different strain rate sensitivities in basal and prismatic slip systems [60] as discussed above. Most dwell fatigue facets have been reported to develop close to hard grain basal planes with the facet normal within about 15° from the loading direction [47]. It is thus interesting to study the strain localisation and load shedding due to the soft grain slip resulting from both basal and prismatic slip systems, respectively.

As shown in Fig. 1(a), the polycrystalline model with a dimension of

Table 1
Slip rule properties for α - β Ti-6242 at 20 °C.^a

Metals	Ti-6242 (Ti-6246) ^b		
Parameters	α -Basal	α -Prismatic	β
ρ_m μm^{-2}		5.0	
ν Hz		1.0×10^{11}	
b μm	2.95×10^{-4}		2.86×10^{-4}
k J K^{-1}		1.381×10^{-23}	
γ'		0.05	
τ_{c0} MPa	270	240	280
ΔF eV	0.4247	0.5621	0.3304
ΔV b^3	1.22	11.97	0.0021

^a These slip properties are based on the studies by Zhang et al. [59], and Zhang et al. [60].

^b The chemical compositions of Ti-6242 and Ti-6246 are close with the exception of the 4% molybdenum content. Physical properties in these two alloys are close [55]. The specific chemical composition of an α - β titanium alloy tends to change the volume fraction of α or β phase [23].

$100 \times 100 \times 6 \mu\text{m}$, meshed by 13,746 ABAQUS C3D20R elements, is deformed in tension to the 0.2% proof stress in 20 s and then held at constant stress to facilitate a two-minute stress dwell at room temperature. Boundary conditions are imposed to give averaged uniaxial deformation and the model polycrystal behaves according to the single crystal plasticity slip rule within the finite element formulation given above in Section 2. Boundary constraint effects potentially influence the analysed fields. However, the analyses carried out nonetheless provide consistent comparisons (with the same overall microstructure and BCs) and importantly, such that the microstructure rankings for dwell sensitivity hold. This is the primary objective in this paper but note that were a detailed assessment of the microstructural quantities to be needed for a fully representative volume element, then additional grains would likely be needed in the analyses. The loading history is shown in Fig. 1(f). The results in Fig. 1, i.e. effective plastic strain $p = \int_0^t \sqrt{\frac{2}{3}} \mathbf{D}^p : \mathbf{D}^p dt$, are taken at the end of the stress dwell. In this particular geometric microstructural model (Fig. 1) the grains shown are taken to be primary α , representative for near- α titanium alloys. α - β microstructures are considered later. The slip properties used are those for the α phase in Ti-6242 in Table 1. The dwell duration is that often adopted in titanium sample tests [16]. The grains of interest are highlighted in Fig. 1(a) as g_0 , g_1 , g_2 and g_3 . Their crystal orientations, as indicated in Fig. 1(b)–(e), are designed to conduct a comparative study of slip accumulation due to basal and prismatic slip, respectively, in a soft grain adjacent to a hard grain badly orientated for slip. All other grains in the model are randomly orientated. It is not unexpected, therefore, to observe a clear 45° shear band developing in the model as the shear stress is maximum along this direction, i.e. for a random texture case, [58]. Note that slip transfer across the grain boundary is not explicitly represented in the present crystal plasticity modelling, but other model formulations such as that of field dislocation mechanics [1] do allow for it. However, the crystal plasticity work on α - β micro-pillars showed the slip development, even across α - β boundaries, to be well captured [59]. Recent direct comparisons between the present crystal plasticity formulation and discrete dislocation plasticity methods for alpha-beta Ti alloys shows that very close agreement is achieved, even local to alpha beta phase boundaries [62].

Firstly, the study is focused on the slip accumulation adjacent to the hard grains. As given in Fig. 1(b) and (c), the grains of interest are illustrated with unit cell orientations in white. The hard grains are labelled g_1 , g_2 , and g_3 (less hard), where g_1 and g_2 are strictly hard as the normal to the basal plane (c-axis) is parallel to the loading direction. In this orientation, typically only pyramidal slip may be activated, but this is difficult since the associated critical resolved shear stress is about three times that of the basal and prism systems [21]. Grain g_3 is less

hard with its c-axis about 10° to the loading direction potentially to allow basal slip when the resolved shear stress is sufficiently high. As shown in Fig. 1, slip is initiated in grain g_0 , and the crystal orientation for g_0 is chosen to enable basal slip activation in Fig. 1(b) and (d) and prismatic activation in Fig. 1(c) and (e), respectively. In addition, they produce the same global Schmid factor of 0.5, and the active plane normal remains identically 45° to the Y axis and slip directions are exactly the same, highlighted by solid white arrows in Fig. 1(b)–(e), so as to introduce a sensible comparison of basal and prism slip accumulation. It is noted that the slip activation and accumulation adjacent to a boundary depends on the crystal orientation combinations and the local stress states developed (through elastic constraint) and the corresponding ‘yield’ strength and rate dependence are also orientation dependent [57], as demonstrated by Savage et al. [45] who reported orientation dependent creep behaviour in Ti-6242 colonies.

Fig. 1(b) shows basal slip activation in grain g_0 which initialized at point 1, and develops towards the soft-hard grain boundary. Here, the basal slip continues to develop from point 2 local to the grain boundary towards point 3. The stress state developed triggers new basal slip within grain g_3 shown in the insert next to Fig. 1(b). Finally, the plastic flow develops back into the soft grain g_0 . In Fig. 1(c), prismatic slip is favoured because of the chosen crystallographic orientation. The development of accumulated slip follows a similar path of 1-2-3-4 as in Fig. 1(b). Comparing Fig. 1(b) and (c), it is evident that there is slightly higher slip accumulation established due to basal slip than for prismatic slip. This reflects recent micropillar experiments and numerical studies, where the intrinsic strain rate sensitivity is higher in the basal slip than for the prismatic slip [60]. This is consistent with the work by Amouzou et al. [2], where the basal slip systems were found to be more rate sensitive than the prismatic in their numerical model in order that the macroscopic response matched the constant strain rate experiments and work hardening tests in their TD and RD textured samples.

Changing the crystallographic orientation of the previously hard grain g_1 to a soft grain as shown in Fig. 1(d) and (e) generates quite different slip fields. Given that slip may now be activated without being inhibited by the presence of a hard grain, a band of slip is seen to develop in a 45° direction, and the plastic strain magnitudes are much greater than that in Fig. 1(b) and (c). The slip development across grains due to the basal slip at grain g_0 , shown in Fig. 1(d), is more significant than that as a result of prism slip in Fig. 1(e). The independent slip system progressive contributions to slip accumulation and band development, represented by the instantaneous effective plastic strain, are shown in Fig. 2.

Fig. 3 shows the loading direction stress, before and after the stress hold in the dwell test, along path A-A' given in Fig. 1(a), and the (soft-hard) grain orientations are shown in Fig. 1(b) and (c), respectively for basal and prismatic slip. Basal slip activation in the soft grain during the stress hold leads to the redistribution of stress from soft g_0 to hard g_1 grains increasing the stress magnitude within the hard grain considerably (by 275 MPa) as shown in Fig. 1(a), which is greater than that which occurs when prism slip is activated in the soft grain for which the redistribution of stress, or load shedding, is shown in Fig. 1(b). In this case, the increase in stress during the dwell is 240 MPa in the hard grain. Hence it is the case that basal slip activation in a soft grain at a soft-hard pair leads to more damaging load shedding than that for prism slip activation, because of the higher intrinsic SRS of basal slip compared to prism slip. A consequence is that dwell fatigue facets may preferentially develop at soft (basal) – hard grain combinations, although soft(prism.)-hard grain pairs may also be possible.

3.2. Response in α - β colony and basketweave structures: Slip transfer and load shedding across the differing phases

The α - β morphology in Ti polycrystals has been demonstrated to affect global rate dependence [57], since the β laths act potentially to inhibit the motion of dislocations and dislocation path. TEM studies

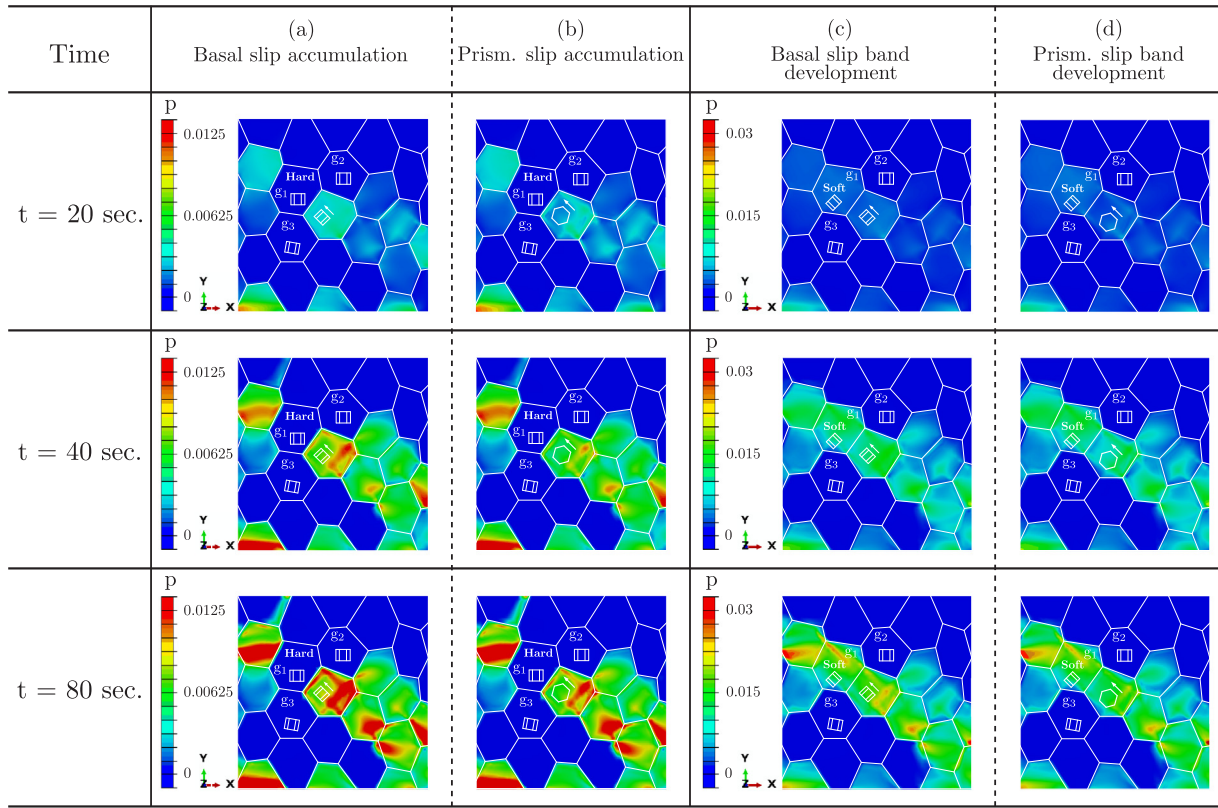


Fig. 2. Snapshots of slip accumulation and slip band development shown in Fig. 1: basal slip accumulation in (a) and (c), and prismatic slip accumulation in (b) and (d), at time $t = 20, 40, 80$ s, respectively. The slip band development in (c) and (d) is represented by the instantaneous effective plastic strain. Note that the microstructures are different, as shown in Fig. 1, in (a) and (c), as well as (b) and (d), respectively.

have been carried out by Suri et al. [52] providing great insight into local deformation mechanisms. The same microstructures were examined by Zhang and Dunne [57] who investigated *microscale kinematic confinement*, owing to polycrystalline heterogeneity, grain boundaries and the BOR relations, and showed that these features can reduce the resulting global rate dependence compared to what would follow from, for example, an unconstrained micro-pillar single crystal response.

Hence here we study the grain, or structural unit level cold creep and dwell behaviour in pure α phase, α - β colony, and basketweave (multiple α variant) structures with differing β thickness. For the grain-level morphological study of creep behaviour, the polycrystal model in Fig. 1(a) is again utilised but with differing morphological manifestations in grain g_3 as shown schematically in Fig. 4(a). Grain g_3 is firstly a pure α grain in Fig. 4(c), but is also then chosen to represent a simple α - β colony structure, a simple four-variant basketweave structure with

thin β laths, and finally a simple four-variant basketweave structure with thick β laths. The β laths modelled are thicker than would sometimes be observed (to facilitate CP modelling) in eg a Widmanstätten structure, but including two thicknesses allows comparisons to be drawn. The four-variant (α_1 to α_4) basketweave orientation combinations are shown schematically in Fig. 4(b), where for simplicity the different habit planes of the α - β interfaces have not been explicitly illustrated. When subjected to stress-controlled dwell loading, the resultant accumulated slip development is shown for each structure in Fig. 4(c) for pure α , Fig. 4(d) for a simple α - β colony structure, Fig. 4(e) for four-variant basketweave with thin β laths and Fig. 4(f) for four-variant basketweave with thick β laths. The thin β lath thickness is selected as $1.6 \mu\text{m}$, within the 0.5 – $2 \mu\text{m}$ range in Ti-6242 colonies [44]. It is doubled to $3.2 \mu\text{m}$ for the thick β laths. What is starkly apparent in Fig. 4 is the dramatic effect of the four-variant basketweave structure

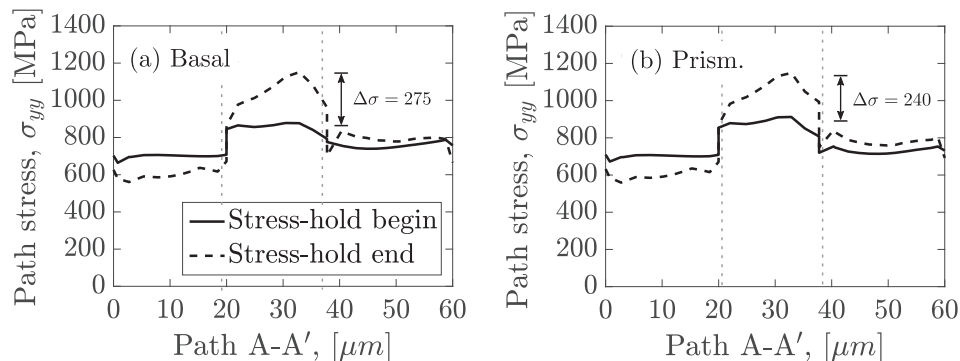


Fig. 3. YY stress along A-A' path in Fig. 1: (a) for basal slip activation in soft grain g_0 in Fig. 1(b), and (b) for prismatic slip activation in Fig. 1(c), respectively. The stress unit is in MPa.

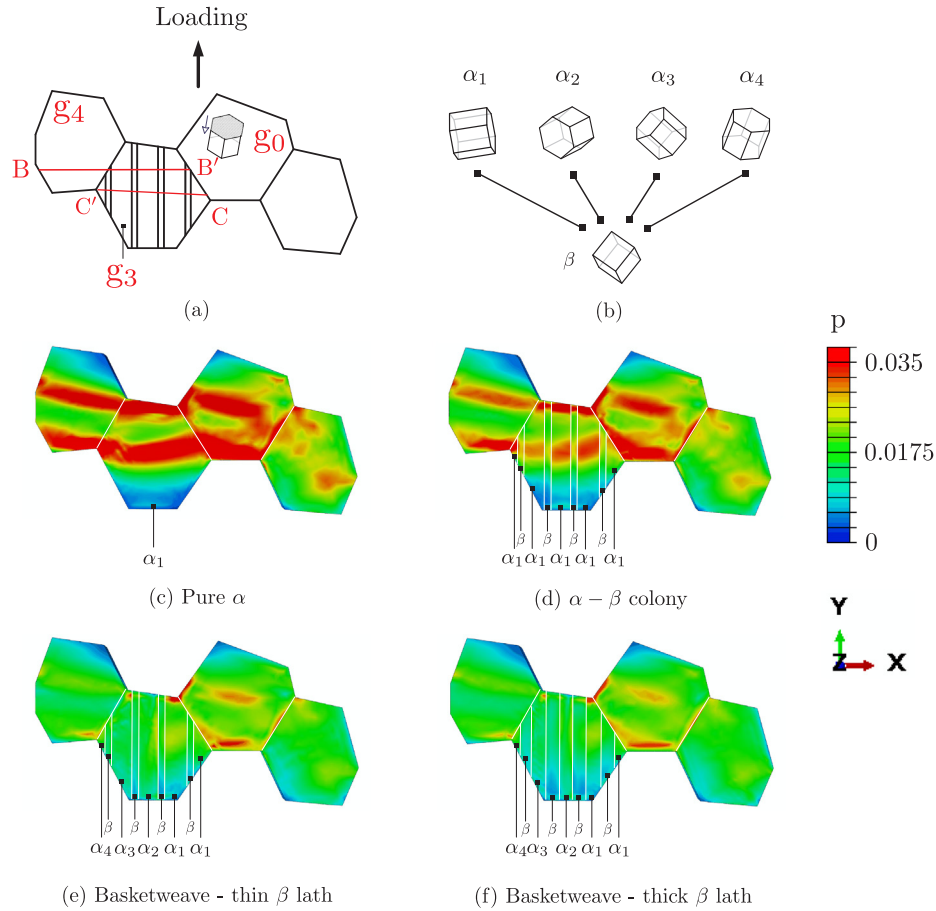


Fig. 4. Slip accumulation during dwell fatigue resulting from differing α - β morphology: (a) region of interest modified from Fig. 1(a), (b) four α -variant BOR crystallography, (c) pure α (g_3) grain, (d) simple α - β colony structure, (e) simple four-variant α - β basketweave structure with thin β laths and (f) simple four-variant α - β basketweave structure with thick β laths.

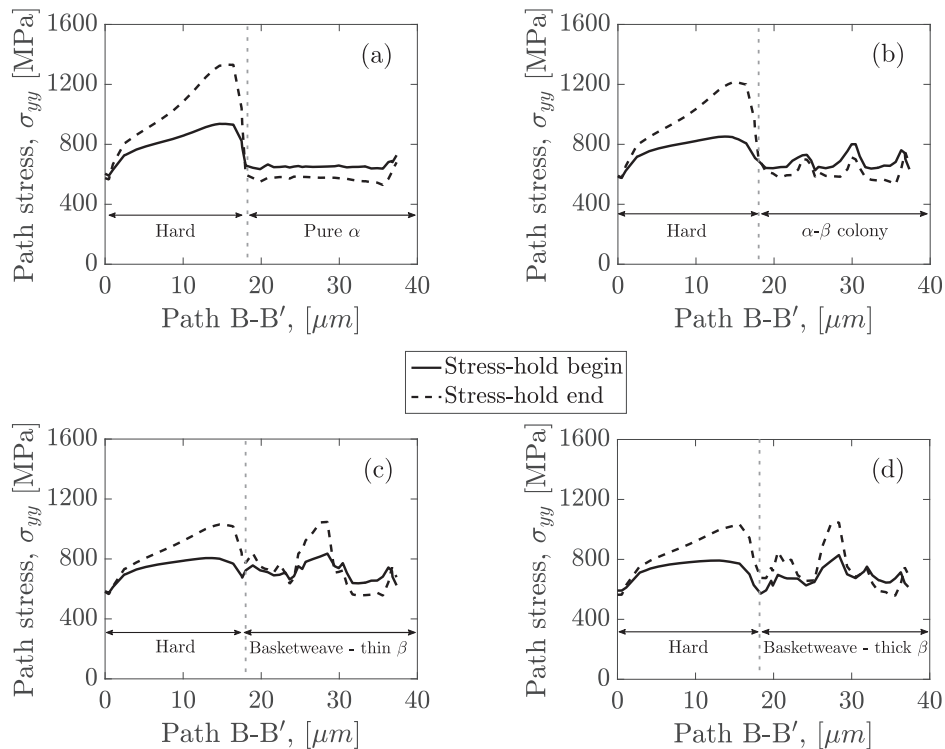


Fig. 5. YY stress along path B-B' (see Fig. 4(a)) in hard-soft grain pair when the soft (g_3) grain is: (a) pure α , (b) α - β colony, (c) four-variant basketweave with thin β laths, and (d) four-variant basketweave with thick β laths, respectively.

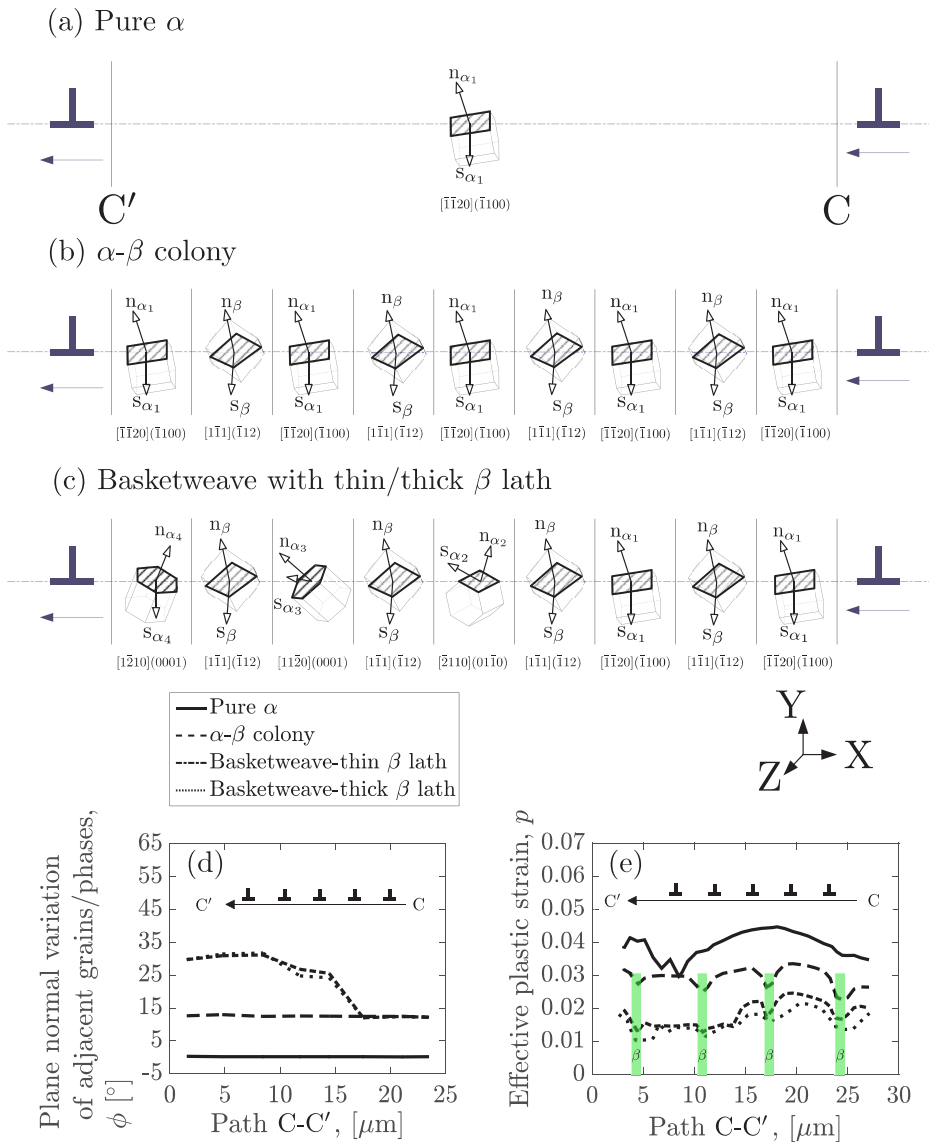


Fig. 6. Path C-C' in grain g_3 in Fig. 4 showing (a) crystallographic orientation for pure α grain, (b) crystallographic orientation distribution for simple α - β colony structure, (c) crystallographic orientation distribution for four-variant basketweave structure, (d) the spatial slip plane normal angular difference variation, and (e) the accumulated slip after a dwell fatigue cycle for the four microstructures indicated. In (a), (b) and (c), the highlighted shaded slip plane is that most favourably orientated to be activated. The thin β lath positions are highlighted in green in (e). (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

((Fig. 4(e) and (f)) on inhibiting slip development during stress dwell as compared to the pure α grain and the simple α - β colony (Fig. 4(c) and (d)). When multiple α variants are present, growing out of the same original β matrix, each α variant tends to rotate differently from the β matrix under deformation though each of them initially remains a semi-coherent phase boundary satisfying the BOR relation to the parent β phase. Multiple variants generate a shorter slip length, interlocking, and inhibition of slip and cold creep accumulation resulting in *micro-slip kinematic confinement*.

It is also very clear that the β lath thickness appears not to play a significant role in inhibiting slip accumulation (Fig. 4(e) and (f)) during dwell such that the over-riding conclusion is that it is the presence of the α variants in the basketweave structures, as opposed to the presence of the β laths, which is inhibiting the dwell effect and slip accumulation (cold creep) in these microstructures. This is consistent with recent findings by Kasemer et al. [25] that the lamellar width has little to no effect on yield strength, macroscopic ductility or hardening in dual phase Ti-6Al-4V alloy. Naturally, the β laths are indirectly important since their presence gives rise to the potential for many α variants, but the creep-inhibiting mechanism is mainly provided by the α variants themselves.

Fig. 5(a) and (b) show the considerable load shedding which takes place during the stress dwell from soft (g_3) structure to adjacent hard

(g_4) grain when the soft g_3 structure is a pure α grain and a simple α - β colony structure. Peak stresses in the adjacent hard grain for these cases exceed 1200 MPa. In clear contrast, however, the stress redistribution or load shedding occurring during the stress dwell for the four-variant basketweave structure (g_3) is significantly reduced leading to a much more uniform spatial stress distribution across the hard-soft grain pairing. Hence the role of the basketweave, and most particularly the α variants, is to inhibit slip accumulation (cold creep) within the α phase during dwell, such that stress redistribution is inhibited and load shedding minimised. This provides the most compelling mechanistic explanation, it is argued, for the observation that the basketweave structures are vastly superior to the colony structures in inhibiting cold creep and hence load shedding, and ultimately in demarcating why the dwell fatigue debit can be large (factor of ~ 10 on life) for *some* Ti-6242 alloys but negligible for *some* Ti-6246 alloys in aero-engine service, depending crucially on their microstructures. The observed behaviour resulted not directly from the chemistry but from the two materials' differing microstructures. This mechanistically explained the recent discrete dislocation work by Zheng et al. [61] in which the homogenised activation energies and volumes led to the inhibition of load shedding in a *particular* Ti-6246 alloy versus marked load shedding in hard-soft grain pairs in a *given* Ti-6242 alloy.

The improved performance of basketweave structures in inhibiting

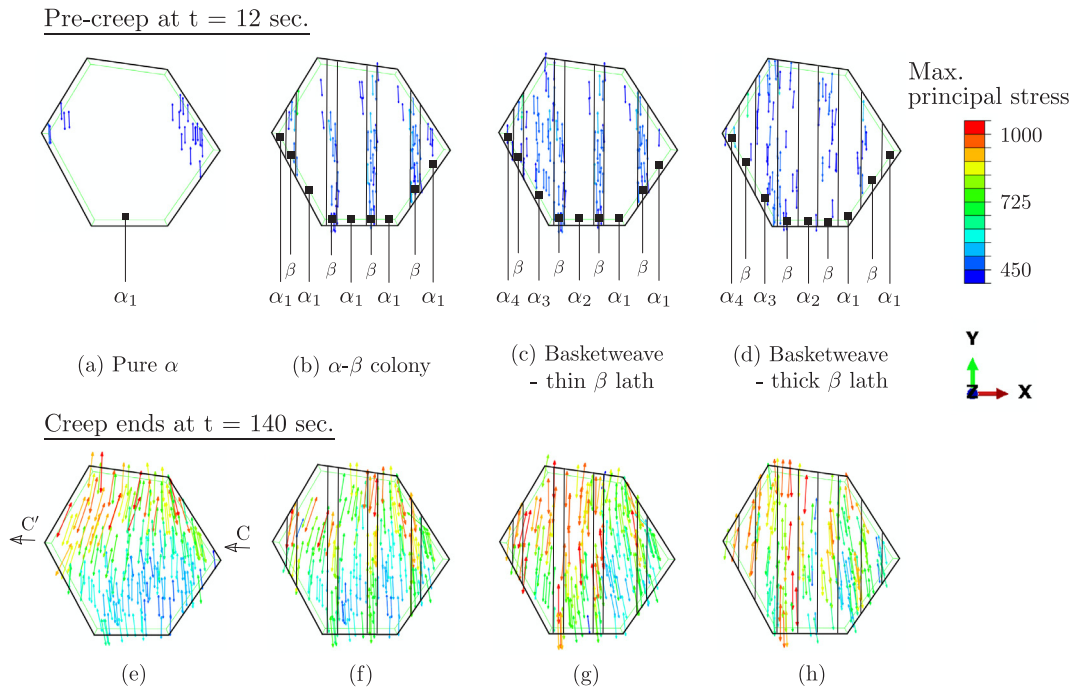


Fig. 7. Microscale max. principal stress distribution in grain g_3 (see Fig. 4(a)) showing results during load-up ($t = 12$ s) in (a)–(d) and at the end of the stress dwell ($t = 140$ s) in (e)–(h) for pure α , α - β colony, and four-variant basketweave structures with thin and thick β laths respectively. The unit of max. principal stress is in

slip lengths and pile-up and the transfer of slip across multiple grains is apparent in Fig. 6. This is addressed in more detail for primary α , colony and basketweave structures in Fig. 6 which shows the crystallographic variation along path C-C' marked in Fig. 4(a) for the three structures considered, namely in Fig. 6(a) pure α , Fig. 6(b) simple α - β colony, and Fig. 6(c) simple four-variant α - β basketweave structure showing the BORs. For primary α in Fig. 6(a), the slip takes place on the same (single) prismatic plane, while in the α - β colony in Fig. 6(b), due to the presence of one (α)-to-one (β) BOR relation, limited inhibition of slip occurs across the α and β phases, as shown in Fig. 6(e). Recent experiments addressing slip transfer across an α - β phase boundary showed that the slip system with the smallest mismatch in plane normal is preferably activated for slip transfer across the phase boundaries [59].

The spatial variation of the slip plane normal angular differences for each of the three structures, defined as $\phi = \arccos(\mathbf{n}_i \cdot \mathbf{n}_j)$, where $\mathbf{n}_i$ and $\mathbf{n}_j$ are plane normals in adjacent slip planes in Fig. 6(b) and (c), are shown in Fig. 6(d). The mismatch angle ϕ , depends on the crystal orientation and the local BOR relation, and may be up to 12° in the Ti-6242 α - β micro pillars. Fig. 6 (d) shows that the plane normal mismatch is about 14° in the present α - β colony illustrated in g_3 of Fig. 6(b).

Since the first four laths along C-C' in the basketweave structure in Fig. 4(e) have the same α_1 - β orientations as in the colony case, the mismatch of plane normal remains the same at 14° . As shown in Fig. 6(d), the first α_2 variant gives a mismatch angle of 25° and other variants give mismatch angle between 25° and 35° . Therefore, the basketweave structure facilitates much more diffuse slip direction development, which is reflected by the more uniform strain distribution for the basketweave structure shown in Fig. 6(e). The accumulated plastic strain during creep remains limited to about 0.02 in the basketweave structures, 0.03 for the colony structures and 0.04 for the primary α in the g_3 grain representations. Importantly, with thicker β lath in the basketweave structure, the creep strain can be decreased further compared to a thinner lath. Note that the initial α variants and β matrix crystal orientations shown in Fig. 6(c) in the basketweave structure are the same for the case of the thin and thick β lath study, and the very small difference shown in Fig. 6(d) arises due to the lattice

rotation update which occurs with deformation. The thicker β lath modifies the local stress redistribution during creep since the high strain rate hardening in the β phase strengthens the β lath and reduces overall strain accumulation.

Here, the heterogeneity of the basketweave α variant orientations is made clear with respect to that for colony and pure α structures. Fig. 6(e) shows the corresponding spatial variation of accumulated slip along path C-C' after dwell fatigue for each of the microstructures demonstrating the dramatic reduction in cold creep in the soft grain (g_3) for the α variant basketweave structures.

A consequence of the dramatic inhibition of cold creep occurring during the stress dwell in the basketweave (g_3) grain, which is well-orientated for α slip, is likely that the stress distributions borne by this grain are different than would be the case if it were pure α or α - β colony. This in turn has consequences for the stress redistribution, or load shedding, on to the adjacent hard grain (i.e. badly orientated for slip).

A further consequence of the α - β morphology is the potential disruption of the local stress states important in nucleating slip. This is investigated through the distributions of local maximum principal stress shown within the soft grain (g_3), and are examined firstly along path C-C' (see Fig. 4(a)) in Fig. 7 for the pre-dwell stage (i.e. during the 20 s load-up) and post-dwell (after the 2 min stress hold) for each of the four microstructures discussed above. The loading history is shown in Fig. 1(f). The arrows indicate the directions of the maximum principal stresses. The distribution density of these arrows indicates the local stress concentration. For instance, localized stresses become more uniform in basketweave structures in Fig. 7(g) and (h), but rather concentrated in the grains in Fig. 7(e) and (f).

There is some indication that the stresses tend to be preferentially carried by the β laths during the load-up shown in Fig. 7(a) to (d), particularly for the colony structure in (b). It is also noticeable that for the basketweave structures in (c) and (d), the stresses are much more uniformly distributed than for either pure α or a colony structured grain, indicating the role of the α variants together with the β laths is to cause a more diffuse and uniform stress distribution in the soft grain. The uniform stress distribution remains the case after the stress hold

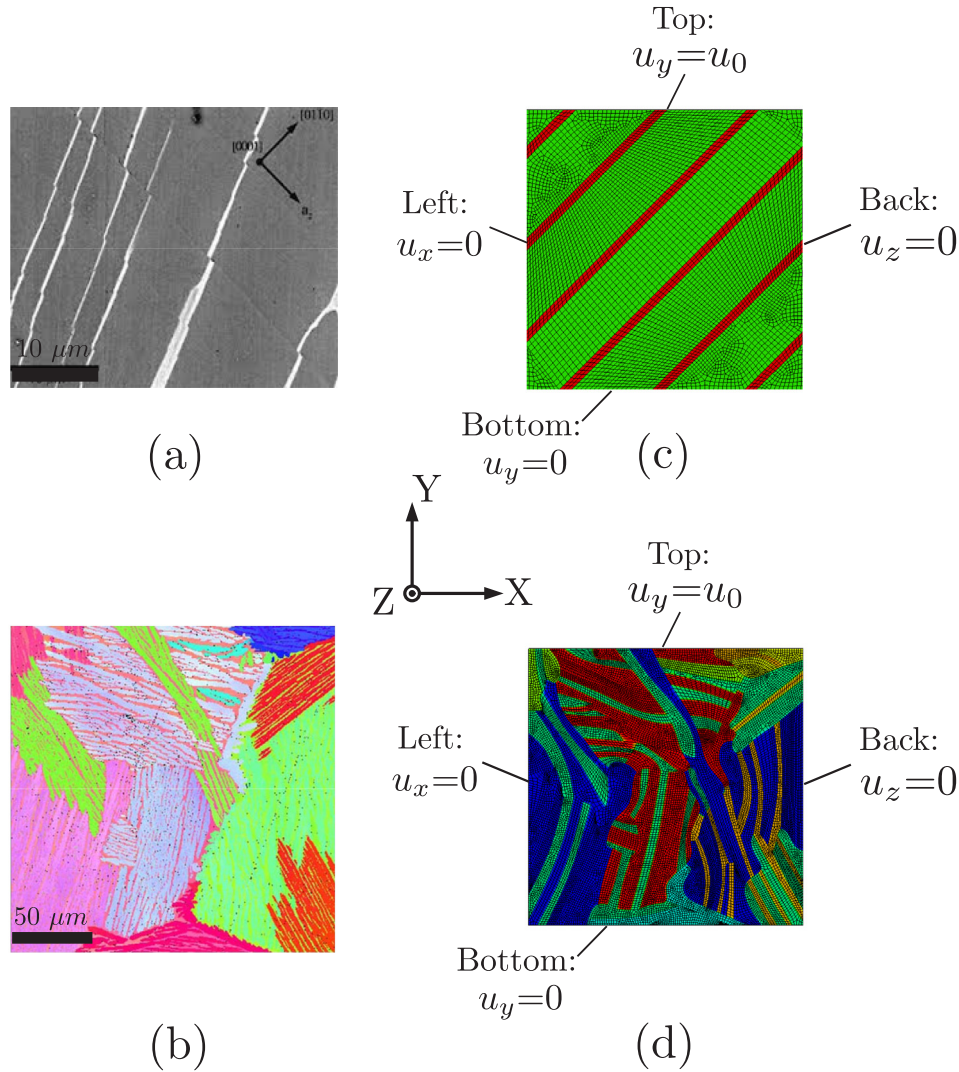


Fig. 8. Microstructural morphology in Ti-6242 and Ti-6246: (a) α - β colony in Ti-6242; (b) basketweave structure in Ti-6246; (c) and (d) are the crystal plasticity finite element representations for (a) and (b), respectively.

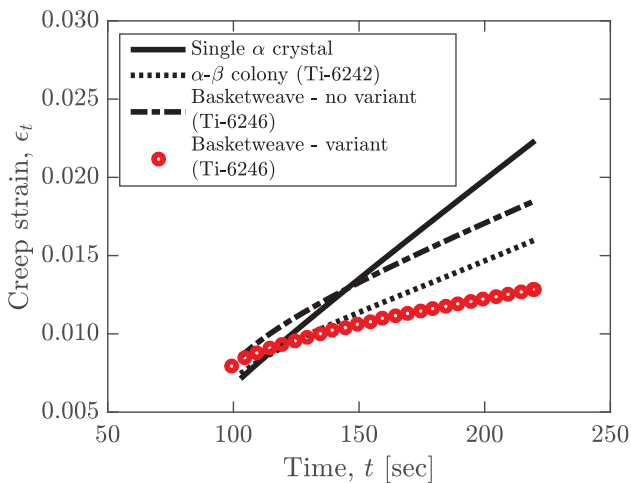


Fig. 9. Macroscale creep strain evolution for polycrystalline models with representative Ti-6242 and Ti-6246 morphologies shown.

(dwell) for the basketweave structures in (g) and (h) particularly but in contrast, much bigger intragranular variations develop in both the pure α grain and the colony structure, suggesting much more significant

stress redistribution into the adjacent grains during the dwell period because of the developing (cold) creep in the soft grain. The stress redistribution is investigated by considering the stresses which develop along path B-B' (shown in Fig. 4(a)) which therefore provides information both for the soft grain (g_3) and an adjacent hard grain (g_4). The results are shown at the beginning and end of the stress hold in Fig. 5 for the range of grain g_3 structures (pure α , α - β colony, four-variant basketweave with thin β laths and four-variant basketweave with thick β laths).

3.3. Polycrystalline study of morphology effects on macroscopic creep during cold dwell loading

The final part of this study examines truly representative morphologies associated with alloys Ti-6242 and Ti-6246 obtained from EBSD characterisation. In order to avoid reproducing figures, readers are referred to the earlier work of Zhang and Dunne [57], where the full geometric details of Ti basketweave structures with and without α variants are described in detail. The α - β colony structures in Ti-6242 and basketweave structures in Ti-6246 are illustrated in Fig. 8(a) and (b), respectively. The cold dwell loading utilised above is applied to these structures, with load-up to 90% of their 0.2% proof stress followed by stress dwell for two minutes to drive cold creep. The material

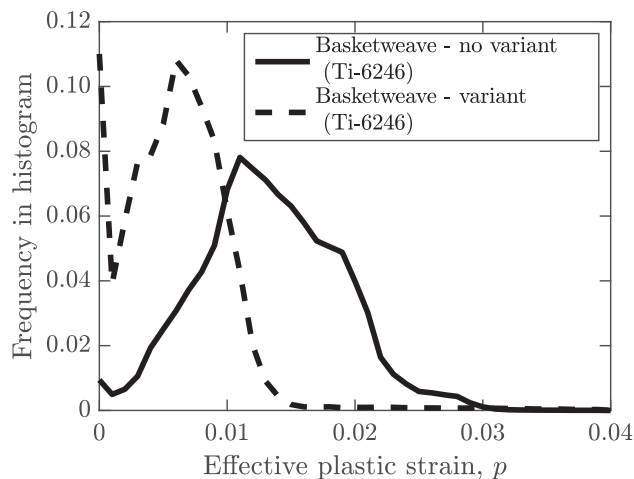


Fig. 10. Distribution of polycrystal plastic strain for basketweave structures with and without α variants.

behaviour is again taken to be that given by the α and β properties in Table 1. Here the macroscopic creep strain, obtained by averaging over the entire polycrystal finite element model, is examined to explore the effect of morphology in cold dwell fatigue resistant material design. Fig. 9 shows the results obtained for the four morphological structures assessed. Entirely consistent with the results obtained above for the model morphological structures, the basketweave structure with most α variants produces the lowest creep strain accumulation. Single phase α polycrystal structures, colony structures, basketweave structures without variants all give higher macroscale creep rates during the stress dwell, and the results confirm that the basketweave structure with multiple α variants leads to considerably reduced creep strain evolution, and that the creep rates obtained for α - β colony structures are in fact close to those for basketweave structures without variants.

A more statistical study is carried out on the impact of the number of variants in the basketweave structures on the creep behaviour. Fig. 10 shows the distribution of effective plastic strains obtained from all points over the full polycrystal plasticity model in order to investigate the full range of strains obtained for the basketweave morphologies in which multiple α variants, labelled as “variant”, and single α variant, labelled as “no variant”, are included in the geometric model. The histogram frequency given in the vertical axis is the number of finite elements normalized by the total number elements in the crystal plasticity model giving rise to the strain levels indicated on the horizontal axis. It is found that the majority of elements in the basketweave structure without variants have developed plastic strains from 0.005 to 0.025. On increasing the α variant number to five in the basketweave structure, referred to as basketweave with variants in Fig. 10, the plastic strains are found to reduce significantly to lie between 0 and 0.01 in most elements. The average macroscale creep strain accumulation is seen to be diminished considerably by the multiple α variants, in effect reducing it by half. The effect of multiple α variants is also to increase the bimodality of the distribution leading to a substantial peak in microstructural locations where the creep strain is confined effectively to near-zero during dwell, in complete contrast to that for the single variant basketweave structure.

This observation is consistent with the grain-level study in Figs. 7(g) and 7(h) in that more α variants generate a much homogenous stress state. This study provides a micromechanistic basis for why (colony) Ti-6242 suffers from substantial dwell debit while (basketweave) Ti-6246 is largely dwell insensitive. In addition, Ti-6246 may have up to 45% β volume fraction within which the α variants grow within larger parent β grains [24]. The total number of α variants may be as high as twelve within a single prior β grain, and multiple α variants in the parent β grain are often observed. The Ti-6242 considered, however, is a typical

near α titanium alloy with less than 15% β phase locating in α grains of about 10–20 μm , such that basketweave structures may have been inhibited, giving its non-optimal dwell performance.

4. Conclusions

The findings from this study are potentially important in guiding choice of material microstructures to achieve better cold creep resistance and dwell fatigue life. They are summarised as follows.

At a hard-soft grain pair under cold dwell loading, basal slip activation in the soft grain gives rise to more damaging stress redistribution, or load shedding on to the hard grain, than that for prism slip because of the basal slip system’s higher intrinsic rate sensitivity.

The α - β colony structures, and basketweave structures with and without Burger Orientation Relationship variants all give rise to differing polycrystal strain rate sensitivities. However, it has been shown that basketweave structures with multiple α variants give the lowest SRSs and that it is the α variants which in the main drive the reduction. β laths within these structures provide the opportunities for multiple, small-scale α variants, hence indirectly contribute to the reduction of SRS. But the primary mechanistic basis is that the α variants inhibit slip accumulation and localisation, and hence stress redistribution and load shedding leading to a much more uniform and diffuse stress distribution across the microstructure. This, it is argued, is the mechanistic basis for why basketweave structures typical of alloy Ti-6246 do not show significant dwell fatigue debit whereas colony structures, often associated with alloy Ti-6242, show a very considerable dwell debit. Hence by careful microstructural design, new dwell-resistant Ti alloys are within reach.

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