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Slip transfer and deformation structures resulting from the low cycle fatigue of near-alpha titanium alloy Ti-6242Si

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ABSTRACT

Near-alpha titanium alloy Ti6242Si, widely used in aero-engine compressor discs, was subjected to low cycle fatigue loading at room temperature. Fracture initiated by facet formation, followed by striated fatigue crack growth prior to final failure. Deformation occurred primarily by planar slip, localized into slip bands in the primary alpha. Within soft-oriented grains in a microtextured region, pile-up of a slip band within one grain resulted in the direct transfer of slip into an adjacent similarly oriented grain. In contrast, pile up of dislocations in a soft grain with a 'hard' oriented neighbour resulted in the activation of few non-connected dislocations in the hard grain, with <a>-type dislocations being activated and the observation of cross-slip. Whilst a high density of dislocations was present from precipitation of secondary alpha in the retained beta ligaments, a little dislocation interaction was observed between the transformed beta and the primary alpha grains.

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1. Introduction

Near- α titanium alloys, such as Ti6242Si, are employed in high temperature aero-engine compressors due to their combination of specific toughness, creep resistance and microstructural stability at temperatures around 400–600 °C (Hall et al., 1973; Tien et al., 1973; Terlinde et al., 1983; Honnorat, 1996). Thus, they are not normally subjected to high stresses while at temperatures that are of concern for dwell fatigue, a phenomenon associated with planar slip and strain-rate sensitive load shedding between grains associated with holds in load of several minutes at <250 °C (Hack and Leverant, 1980; Tal-Gutelmacher and Eliezer, 2005; Tan et al., 2015). During a flight cycle, a component is exposed to a variety of loading regimes. High stresses in the low cycle fatigue regime may be experienced at the thrust peak associated with take-off, and for the rotor bore this may be at lower temperatures if long soak times on taxi are reduced in an attempt to reduce local ground level airport emissions, which can be problematic at high volume locations such as Heathrow in the UK. Such low temperature dwells might give a cause for concern regarding dwell fatigue. In contrast, other periods of high throttle setting will be at elevated temperatures where dwell effects are absent. These effects continue to compromise design (McBagonluri et al., 2005; Qiu et al., 2014) and hence are very important to understand. In this work, the low cycle fatigue behaviour of the alloy is investigated.

The fatigue behaviour of near- α alloys mainly depends on microstructural variables such as volume fraction of primary alpha (Shen et al., 2004), its morphology (Eylon and Hall, 1977; Kassner et al., 1999), whether the variant state is colony or

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basketweave (Evans, 1987; Sinha et al., 2006a; Dunne et al., 2007a; Dunne and Rugg, 2008) and micro- and macro-texture (Evans and Bache, 1994; Bache et al., 1997a; Bache, 2003; Dunne et al., 2007b; Pilchak, 2014). Furthermore, it is believed that hard/soft grain pairs act as fatigue initiation sites (Dunne et al., 2007a, 2007b; Dunne and Rugg, 2008) and this occurs by the formation of near-{0002} facets near-perpendicular to the loading direction on grains poorly oriented for $\langle a \rangle$ slip ('hard grains'), due to load shedding by adjacent grains that are well oriented for prism $\langle a \rangle$ slip ('soft grains') (Bache et al., 1997b; Bache and Evans, 2003; Kirane and Ghosh, 2008; Anahid et al., 2011; Zheng et al., 2016a,b). However, these facets may, or may not, show evidence of plasticity, and are believed to grow over more than one fatigue cycle – i.e. they are not formed simply by cleavage.

Ti6242Si (Ti-6Al-2Sn-4Zr-2Mo-0.1Si, wt.%) is a two-phase alloy with soft α phase and regions of retained β phase containing fine scale secondary α , where plasticity is assumed to initiate in the primary alpha (α_p) grains. Dislocation generation and transport within a grain can then result (Lee et al., 1990a, 1990b) in: (1) incorporation of the piled-up dislocations in the grain boundary and subsequent decomposition into grain boundary dislocations, (2) direct transfer of piled up dislocations through a grain boundary into the neighbouring grain, (3) piled up dislocations slipping along the grain boundary plane, (4) transfer of a dislocation through the boundary with a residual dislocation left at the grain boundary and/or (5) dislocations being ejected back into the pile-up grain. These processes involve the dynamics of dislocation motion as well as the geometry of slip at grain boundaries (Lee et al., 1990a). Slip transfer across the grain boundary occurs when the following three conditions (Lee et al., 1990a) are satisfied: (1) The angle between the lines of intersection of the incoming and outgoing slip planes with the grain boundary plane should be minimized; (2) the resolved shear stress acting on the possible slip system in the adjoining grain should be maximized; (3) the magnitude of the Burgers vector of any grain boundary dislocation produced during slip transfer should be minimized. Other stress relieving mechanisms are possible when the slip systems of lower critical stress are not favourable.

Slip transfer is connected to fatigue behaviour insofar as a lack of slip transfer will be associated with the production of sufficient stress concentration to result in the initiation of a fatigue crack in an initiating grain; conversely, easy slip transfer without the production of interface defects and debris will result in homogenous plastic deformation. Therefore, micro- and macro-texture will influence fatigue crack initiation. Of importance will also be the micromechanical loading state – the compatibility of the elastic moduli and plasticity between the grains, and therefore the stress perturbations associated with discontinuities such as grain boundaries (Hasija et al., 2003; Dunne et al., 2007a; Venkatramani et al., 2007; Dunne and Rugg, 2008). Macrozones, that is, regions of common orientation inherited from the ~0.5 mm prior beta grains, have been particularly implicated in fatigue performance (Le Biavant et al., 2002; Germain et al., 2005, 2008; Gey et al., 2012), and there has been discussion that macrozones may act as large structural units that deform simultaneously (Evans and Bache, 1994; Bache et al., 1997a; Bache, 2003), with fatigue crack initiation occurring at the interface between hard-oriented and soft-oriented macrozones (Bantounas et al., 2009; Tympel et al., 2016).

In this work, the dislocation structures produced during low cycle fatigue loading in Ti6246Si are investigated in detail, using transmission electron microscopy, with the aim of elucidating how slip transfer and grain boundary interactions more generally may contribute to fatigue crack initiation and propagation. Ti-6242Si was chosen for study because it is a near-alpha alloy that can, in certain microstructural conditions and loading regimes, suffer from cold dwell fatigue and which has a relatively simple equiaxed primary alpha microstructure in a transformed β matrix consisting of secondary alpha laths separated by a thin layer of β phase.

2. Experimental description

Ti6242Si buttons with composition shown in Table 1 were produced by arc melting in low pressure Ar from high purity Ti sponge, elemental additions and TiO powder. Small amounts of Si and oxygen are added to the alloy for high temperature properties and strength respectively. The alloy was processed by rolling in both the β and $\alpha + \beta$ domains, recrystallized in the $\alpha + \beta$ domain at 950 °C for 5 h and air-cooled. The alloy was then aged at 593 °C for 8 h and air cooled to promote Ti₃Al precipitation. A recrystallization temperature of 45 °C below T_β was chosen in order to obtain a high volume fraction of primary α .

Mechanical tests were conducted using Mayes servohydraulic machines with Instron 8800 controllers. Static tensile tests were carried out on dog bone shaped samples at a strain rate of 10^{-3} s^{-1} . Low cycle fatigue (LCF) tests were carried out on cylindrical samples with a 2.9 mm diameter and a 15 mm long gauge length. The tests were performed using a trapezoidal waveform with a ramp up/down time of 1 s, a 1 s hold at maximum stress of 95% of yield stress and an R ratio of 0.05.

A Zeiss Auriga field emission gun scanning electron microscope (FEG-SEM) in secondary electron imaging mode was used for metallography and fractography. Electron back-scattered diffraction (EBSD) was used for orientation imaging microscopy

Table 1

Nominal chemical composition of the alloy (wt.%).

Element	Al	Zr	Sn	Mo	Si	Fe	O	Y	Ti
Wt.%	6.0	4.0	2.0	2.0	0.1	0.25	0.15	0.005	Balance

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