Influence of Complex LCF and Dwell Load Regimes on Fatigue of Ti-6Al-4V

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Abstract

Real components are usually subjected to variable amplitude fatigue, and yet the deformation micromechanisms that occur due to such load changes have barely been the subject of study. Here, unidirectionally rolled equiaxed Ti-6Al-4V plate was subjected to mixed dwell and variable amplitude low cycle fatigue (LCF), with the finding that overloads near the yield stress were found to retard subsequent fatigue crack growth, whilst elastic underloads were found to accelerate subsequent growth. Dwell intervals were found to be especially damaging, to a far greater extent than either dwell or LCF alone. Dwell facets were found to initiate subsurface and to be smoother than LCF facets, but were otherwise similar in orientation ($\sim 30^\circ$ to the loading axis) and crystallographic plane, $2\text{--}13^\circ$ from (0002). However, no alteration of the slip bands underlying striations was observed at the point of load changes using TEM. In failure investigation, striation counting is an important tool; the loading changes used were not found to affect the number of striations formed. Dislocation networks were found between similarly oriented grains in the as-received material, which disintegrated under dwell loading and at high stresses.

Keywords: Titanium alloys, Fatigue, Facets, TEM, Dislocation structures

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1. Introduction

Titanium alloys are widely used in aerospace gas turbines at temperatures up to around 600°C owing to their superior specific fatigue strength, especially for rotating parts such as rotor discs and blades [1]. In service such components
5 undergo complex loading conditions. For example during takeoff, the engine experiences a period of high spool speeds, potentially leading to the cold dwell fatigue failure regime [2, 3]. In other parts of the flight cycle, there are other changes in loading regime and temperature, such at climb and cruise. Therefore, there is a strong practical motivation to understand how variable amplitude
10 loading affects fatigue crack initiation and growth.

More than 85% of the lifetime of Ti-6Al-4V subject to cyclic loading can be attributed to fatigue crack initiation [4]¹. Conventionally, initiation is taken to occur at slip bands, which are continuously cycled during fatigue. Hydrogen is suspected to interact with dislocations, increasing the energy for cross slip,
15 promoting slip bands [5]. High stress ratios (R) favour initiation at the sample surface with the formation of initiation facets. The initiation facets were found to lie around 10° from the basal plane. Surface facets were tilted 30°, whilst subsurface facets were 30° – 60° towards the loading direction [6]. The point of fracture origin is susceptible to the loading conditions and can also occur
20 subsurface [7].

Individual cycles can leave single growth features on the fracture surface which are termed striations. During crack opening and closing, slip systems at the crack tip are activated [8, 9]. This slip activity leads to decohesion,

¹It should be noted that [4] defines initiation pragmatically, at a crack length of 0.5 mm. For the purposes of the present work, we prefer to define initiation microstructurally, as the formation of a single cracked α_p grain, of approximate dimension 10 μm . Using standard da/dN curves for Ti-6Al-4V, growing a crack between these two sizes would take $\sim 20,000$ cycles at a stress level of 820 MPa, and $\sim 14,000$ cycles at 880 MPa. Therefore ‘microstructural’ crack initiation can be inferred to require a significant fraction of life at lower stress levels, but may occur quite quickly near the yield stress.

progressing the crack by an increment Δa . Striation-like features were found
25 to be created by two bisecting prismatic planes [10]. A single striation can be
formed by a single cycle, within a particular range of ΔK . For example, below
 $\sim 10 \text{ MPa} \sqrt{\text{m}}$ the plastic zone size was found to be too small for striation
formation [11]. Above $\sim 50 \text{ MPa} \sqrt{\text{m}}$ the crack progression per cycle was larger
than the corresponding striation separation on the fracture surface.

30 Cold dwell fatigue occurs at near-ambient temperature where ($\sim 120 \text{ s}$) holds
at load lead to large reductions in cyclic life, at stresses significantly below the
yield stress. Crack nucleation is usually subsurface, with $\{0002\}$ dwell initia-
tion facets found perpendicular to the loading axis. The $\{0002\}$ direction is
the stiffest orientation, but a grain in this orientation can only slip by $\langle c + a \rangle$
35 slip, which is hard to activate. The modified Stroh model for dwell facet forma-
tion [2, 12] suggests that deformation in this ‘hard’ grain is initiated by a dis-
location pile-up in an adjacent grain that is well oriented for $\langle a \rangle$ -slip. However,
Brandes et al.[13] found dwell sensitivity in samples without a microstructure
that had the required ‘hard’ to ‘soft’ grain combinations. Crystal plasticity fi-
40 nite element modelling has shown that the time dependence of the phenomenon
is also important [14].

Relatively little work has been done that links these micromechanical obser-
vations to the behaviour observed in variable amplitude loading regimes. For
example, after a large plastic overload, a decrement in the fatigue crack growth
45 rate has been observed, which was attributed to shielding by the crack tip plas-
tic zone [15, 16, 17, 18]. However, evidence for this proposition has not been
advanced from TEM observations, most likely because, until recently, it was
very difficult to extract TEM foils from fracture surfaces. In addition, most
such studies examined long cracks grown from notches (da/dN vs. ΔK), and
50 so the applicability of these concepts to smaller overloads, or even to intervals
of reduced loading, has yet to be established. Finally, the effect on short crack
initiation and growth has yet to be studied.

In the present study we attempt to address these gaps in our understand-
ing by examining the effect of relatively small load changes, and of transitions

55 from low cycle to dwell fatigue loading, on the fatigue behaviour of un-notched, naturally initiated cracks in Ti-6Al-4V. After testing, the fracture surfaces were examined and the underlying dislocation structures observed in both initiation facets and the striations occurring during load changes.

2. Experimental Procedure

60 Unidirectionally rolled Ti-6Al-4V was employed, which had been forged in the β field, rolled in the $\alpha + \beta$ and then creep flattened. The microstructure consisted of a near equiaxed α -Ti with an average grain size of 10.1 μm , Figure 1a. An EBSD map of the cross section can be seen in Figure 2. Macrozones of similarly oriented α -Ti grains were found, extending in the rolling plane and up to 150 μm thick in the normal direction (ND). The macrozone texture was identified as $\{11\bar{2}0\} \langle 0002 \rangle$ [19]. Grains within that zone have the c-axis aligned with the transverse direction (TD) of the material. Grains outside these macrozones have a $\{10\bar{1}0\} 0002$ texture with the $\langle 0002 \rangle$ plane normal in the rolling direction. Threaded cylindrical samples were used with a 2.5 mm gauge
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70 diameter, and the loading direction parallel to the TD. The samples were surface ground after turning in the lathe, and then hand-polished using colloidal silica in an attempt to minimise the effects of surface roughness and any residual stresses introduced by turning.

[Figure 1 about here.]

75 [Figure 2 about here.]

A series of baseline low cycle fatigue (LCF) tests were conducted using a 100 kN servohydraulic testing machine, along with tests employing alternating blocks at the same loads, Figure 1b. A 1-1-1-1 loading cycle was utilised in load control, loading in 1 s, holding at stresses of 820 MPa, 880 MPa and 940 MPa
80 for 1 s (2 min for dwell), then unloading in 1 s, and holding at 20 MPa for 1s. The variable amplitude block loading tests were combinations of LCF 880 MPa with 820 MPa, 940 MPa and dwell, Table 1.

Fracture surfaces were observed using a Zeiss Auriga dual beam FIB-FEGSEM with an integrated Bruker EBSD camera. Dislocation analysis was performed
85 with a JEOL 2000FX 200 kV TEM. Site specific foils of surface features were prepared by lift out using a dual-beam Helios NanoLab 600 FIB-FEGSEM. Non-site specific foils were prepared by thinning and electro polishing with a solution of 3% Perchloric acid, 57% Methanol and 40% Butan-1-ol at $-50^{\circ}C$.

Quantitative tilt fractography (QTF) was used to calculate the spatial ori-
90 entation of features on the fracture surface, as described by Sinha et al. [20]. Two SEM images at the same magnification, but two different stage tilts, are made. In both images four reference points, three forming the feature plane and one as coordinate origin, must be identified. The coordinates of the reference points are then used to calculate the plane normal. When the chosen reference
95 point acting as origin is alternated, 4 different planes of the facet can be calculated, giving information about the shape of the fracture feature. The standard deviation of the 4 planes was used as an indication of the degree of curvature of the features.

3. Results

100 3.1. Fracture overview

[Table 1 about here.]

The failure endurance and fractographic behaviour are summarised in Table 1. Repeat testing of the 820 and 880 MPa LCF samples gave lives of 129,500
and 25,836 cycles, respectively. The endurance at 820 MPa, 880 MPa and 940
105 MPa was approximately 280k, 22k, and 10k cycles, respectively. Dwell fatigue loading at 880 MPa resulted in a life that was only $\sim 40\%$ of the LCF life at the same stress. The combined block loading samples were not found to obey Miner's rule especially well [9]. The application of overloads in test Z (LCF 880/940, 100A/10C), resulted in 21% greater life than if conditions A and C
110 were applied alone; $(21642/21843 + 2160/9613) = C_{Miner} = 1.21$. Similarly, blocks of reduced loading relative to the reference stress of 880 MPa in test Y

were found to be damaging, giving $C_{Miner} = 0.76$, whilst the periodic dwells in test X were also damaging, $C_{Miner} = 0.42$.

When tested at stresses below 940 MPa, the crack leading to failure initiated
 115 at the surface, whilst at 940 MPa the initiating crack was subsurface. The
 combination of different LCF block loads always resulted in a failure by surface
 initiation. Under dwell, at least 9 subsurface initiation sites were identified.
 In combination with LCF loads, over twice as many subsurface initiation sites
 could be observed. As seen in the overview SEM images in Figure 3, subsurface
 120 initiated cracks grow along the rolling direction, forming needle-like features on
 the fracture surface. Those features varied in length as well as in their position
 along the gauge length. The largest on the Dwell 880 sample (D) was 1.35 mm
 long. On the LCF/Dwell sample (X), two large propagated cracks were found,
 having a length of 1.16 mm and 0.86 mm. In contrast, surface initiated cracks
 125 grew semi-elliptically towards the sample centre.

[Figure 3 about here.]

3.2. Fracture initiation facets

[Figure 4 about here.]

The initiation sites of the major cracks are shown in Figure 4. Generally,
 130 smooth initiating facets were observed, except for the two higher stress LCF
 tests at 880 and 940 MPa (A and C). In those two cases, the initiation sites
 were torn, curved and showed crack arrest marks; nucleation appeared to have
 occurred at a surface asperity. In general, the initiating facets were similar in
 size to the diameter of the α -Ti grains, suggesting transgranular fracture along
 135 a crystallographic plane. In the LCF 880/940 sample (Z), slip traces could
 be observed but in the other cases, the initiation facets were smooth. The
 subsurface facets appear smoother with less curvature than surface facets. The
 dwell 880 facet was the only one that showed no signs of plastic deformation;
 we suggest that the step observed was due to an adjacent, similarly-oriented α
 140 grain.

Surface initiated cracks also formed growth facets at distances up to 150 μm from the initiation site. Growth facets were found to be more structured, with ridges, step-to-step complementary and non-complementary features. The plane of fracture was measured by means of TEM foil liftouts and a combination
145 of quantitative tilt fractography (QTF) with electron backscatter diffraction (EBSD).

The facet normal and (0002) pole of the facets with respect to the sample coordinate system are displayed in a stereographic projection in Figure 5. The point $(0^\circ, 90^\circ)$ is the loading direction, coincident with the transverse direction
150 (TD) of the rolled plate material. The x -axis is the angle between the facet normal (or (0002) pole) towards the rolling direction when projected on a plane orthogonal to the normal direction. The coordinates $(90^\circ, 0^\circ)$ corresponds to the rolling direction. The y -axis is the angle between the facet normal (or (0002) pole) to the normal direction. The normal direction is given by the
155 coordinates $(0^\circ, 0^\circ)$. The spatial orientation of the facet is shown as a green cross, with the size of the cross corresponding to the variation in facet normal (small indicates a smooth, planar facet). The colour coded contours show the (0002) pole measured using EBSD. To visualise the amount of misorientation between fracture plane and basal pole, 3 circles of 5° , 10° and 15° misorientation
160 are drawn around the fracture plane. It should be pointed out to the reader that although this projection shows the same features as in Figure 4, those SEM images were not taken with the rolling and normal directions parallel to the microscope axis.

[Figure 5 about here.]

165 None of the initiation sites were found to be perpendicular to the loading direction. On average, the surface and subsurface initiation sites were found at inclinations of 25.9° and 34.5° , respectively. Furthermore, the subsurface facets were flatter than the surface ones, indicated by an extremely small green cross. The contour lines of the basal poles of the facets show two distinct peaks in
170 several samples. Consequently, a single facet can consist of more than just one

α_p grain. The average misorientation between the (0002) pole and the plane of fracture was found to be 6.4° in surface facets and 5.2° in subsurface facets. For selected facets, TEM foils were lifted out and the misorientation between (0002) plane and fracture plane measured. The results compared to QTF/EBSD are presented in Table 2. The misorientations measured with TEM correlate well with the data from QTF/EBSD. However, the TEM technique is only able to acquire data for the specific site from which foil was taken; in cases where a facet was formed by more than one grain, TEM data from only one grain was acquired.

[Table 2 about here.]

3.3. The spatial orientations of initiation and propagation features

A comparison between the spatial orientation of crack initiation and propagation features of the samples loaded to 880 MPa (A,D,X) can be seen in Figure 6. Propagation facets formed by LCF fatigue showed the largest average tilt from the loading direction of 38.1° , followed by Dwell/LCF 880 initiation (33.2°), Dwell 880 initiation (31.4°) and striated grains (20.8°). Strikingly, propagation facets in LCF showed no preferred orientation, except that they tended to face towards the crack initiation site. In contrast, the striated surface normals tended to cluster in the ND-TD plane, close to the TD - which was the loading axis. It may be significant that, for the $\{0002\} \parallel \text{RD}$ macrozones, these striation orientations are consistent with the available prism $\langle a \rangle$ slip planes. Dwell initiation facets were near-TD and (0002), but inclined towards the RD. However, in combined LCF/dwell these initiation facets showed a much greater spread of orientations.

[Figure 6 about here.]

3.4. Striated crack growth

The formation of fatigue striations was observed in samples with an LCF component. In purely LCF loaded samples, at a distance of 60-180 μm from the

initiation, striations began to be observed. The striation separation progressively increased until the fracture. The sample that was loaded purely in dwell
200 did not show striations as a relevant growth feature. In combination with LCF cycling, striations were found in macrozones adjacent (presumably $\{0002\} \parallel$ RD) to the main faceted crack growth feature. The total interval over which striations could be observed in that sample was approximately $200 \mu\text{m}$, which
205 was only $\approx 25\%$ of the length observed in samples with no dwell.

Under the assumption that one striation is formed during one loading cycle, they were used as a tool for measuring crack growth rates, post mortem, such that the crack propagation rate (da/dN) is equated to the measured striation separation. This assumption may not always be warranted, *e.g.* in situations
210 where striations are not visible, at very short crack growth rates, or for other reasons. The stress intensity ΔK was calculated according to the Astiz model presented by Toribio et al. [21]. The Astiz model is specific for a circular cross section and takes into account the width to depth ratio of the crack. This striation counting process was straightforward for samples with only one
215 continuous fatigue stress. However, samples which underwent block loads at different stresses, the striation separation alternated, as seen in Figure 7. A change in load resulted in a visible change in striation separation, which allowed the load to be related to the striation separation at a given point in the fracture surface. For the particular grain seen in Figure 7, a change in stress from 820
220 MPa to 880 MPa caused the separation to increase.

[Figure 7 about here.]

The result of the striation count can be seen in Figure 8. Each graph compares a continuously cycled sample with the corresponding same stress block under complex loading. As seen by the wider 95% confidence interval the scatter of the separation is larger under combined block loading. For the 880 MPa
225 cycles the rate of crack propagation seemed to have reduced by the presence of 940 MPa overloads and increased when underloads of 820 MPa were applied. The rate of crack propagation during the underloads themselves seemed not to

change due to the presence of the 880 MPa block loads. For the 940 MPa over-
 230 loads a small reduction in propagation speed was found under complex loading,
 but within the fitted confidence band. The slope of the linear regressions only
 changed in the sample with combined 880/820 MPa loading (Y), during the
 880 MPa loading portion; here the slope increased slightly, indicating increased
 crack acceleration.

235 [Figure 8 about here.]

3.5. Dislocation networks

Apart from fractography, the focus of this work lies in the dislocation anal-
 ysis of the fracture initiation sites and propagation features. The as-received
 material showed mainly $\langle a \rangle$ -type dislocations on prismatic and pyramidal slip
 240 planes. Dislocation network structures such as the one seen in Figure 9a were
 observed in multiple grains. $\mathbf{g} \cdot \mathbf{b}$ and trace analysis showed that three disloca-
 tions periodically form nodes, forming a net. The dislocations were identified as
 $(0\bar{1}10)[2\bar{1}\bar{1}0]$, $(1\bar{1}01)[11\bar{2}0]$ and $(10\bar{1}1)[1\bar{2}10]$ screw dislocations. Networks were
 found in grains with different orientations and are easy to identify, making them
 245 a good structure to investigate the effect of different fatigue loads.

[Figure 9 about here.]

The deformation of the dislocation network under continuous cycling and
 dwell loads was investigated by preparing TEM foils from the gauge section,
 several mm from the fracture surface. The LCF 820 MPa sample (B) was se-
 250 lected as a representative of continuous loading, and compared with the dwell
 880 MPa sample (Z). As seen in Figure 9b, the network shows no sign of defor-
 mation after 820 MPa cyclic fatigue. The network was again found to be formed
 by $\langle a \rangle$ -type screw dislocations on pyramidal and prismatic planes. A QTF study
 showed that the networks lay close to the $\{1\bar{1}02\}$ plane, with a similar disloca-
 255 tion density to that of the as received material. Dislocation aggregation at the
 grain boundary indicated a degree of plastic deformation in the gauge section.

In strong contrast to this is the network after dwell at 880 MPa as seen in Figure 9c-d. The network of $\langle a \rangle$ -type screw dislocations was less homogeneous in cell size and showed distortion and defects around the edges. Also, the network was surrounded by paired dislocations (double lines) with darker contrast. The paired dislocations were found to be of $\langle a + c \rangle$ -type lying on pyramidal slip planes. They showed evidence of cross slip between pyramidal planes and were observed near the edge of the network.

3.6. Dislocation structure within initiation facets

The results of the dislocation study of the subsurface facet causing the LCF 940 MPa sample (D) to fail are shown in Figure 10. An overview of the facet forming grain revealed slip bands of $\langle a \rangle$ -type screw character on the $(0\bar{1}10)$, $(10\bar{1}0)$ and $(10\bar{1}1)$ planes, (a). Near the facet surface, dislocations identified as $\langle a \rangle$ -type edge or $\langle c + a \rangle$ -type were found (b). Most of these dislocations were curved, but overall they were found to follow the trace of the basal plane. It can be clearly seen that these change to different basal slip planes by forming jogs. Below the surface a heavily distorted network can be seen (d). The network was almost completely untangled and only a few irregular sized cells remained.

[Figure 10 about here.]

The formation of slip bands could not be observed in the subsurface initiation facet found in the LCF 820 MPa sample (B), Figure 11a. Near the surface, curved edge dislocations with an $\langle a \rangle$ -type Burgers vector were found. The edge dislocations were almost exclusively found at the fracture surface near a grain boundary (b & c). Below the fracture surface, undeformed and deformed networks were visible (d). Dislocation analysis on surface initiation facets was also attempted. However, the dislocation density was found to be so high that it was impossible to identify them individually. Furthermore, the grain showed so much deformation that it was impossible to have the same diffraction condition over the whole grain.

[Figure 11 about here.]

3.7. The dislocation structure in striated, block-loaded grains

To investigate the dislocation mechanisms involved during changes in loading regime, TEM foils of striated grains from the LCF 880/820 sample (Y) and LCF/Dwell 880 sample (X) were prepared, Figure 12. The foils were lifted from striated grains which showed a change in striation separation caused by a load change. The TEM foil plane was chosen to be perpendicular to the striations. Clearly visible in both samples were slip bands on two pyramidal planes. At the point where the load regime changes, no significant change of slip could be observed. However, clearly visible was a deflection of the crack at this point. On the surface of the LCF 880/820 sample, Figure 12a, the load step up to 880 MPa created a depression. The crack path returned to its original direction before the next step down to 820 MPa. Similarly, in the LCF/Dwell 880 MPa sample (X) no change in slip could be seen, but the 2 minute dwell period caused a large step on the fracture surface. The step height was found to be $0.56\text{ }\mu\text{m}$, which was more than an order of magnitude greater than the height of LCF striations in that grain.

[Figure 12 about here.]

4. Discussion

4.1. Facet crystallography

The QTF study on facets has shown distinct differences between dwell and LCF facets formed during crack initiation or propagation. The first obvious difference was that dwell initiation facets were generally subsurface whilst LCF initiation facets were at the surface. The observed tilts were higher in the subsurface than in the surface facets. During crack propagation this order changed. LCF propagation facets tilted further and dwell propagation facets tilted further while dwell propagation facets are reported to exhibit a smaller tilt of $\sim 10^\circ$ [20, 22]. The dimensions and orientation of dwell propagation facets suggested that they formed in a macrozone. The small inclination of the facet normal to

loading direction caused the macrozone to glint, which was visible to the naked
315 eye. Contrary to this were the observations of Sinha et al. [22, 20]. There, facets
formed by continuous cycling were reported to be almost normal to the loading
direction, while dwell fatigue facets showed small tilt of $\sim 15^\circ$. The fact that
Sinha did not distinguish between initiation or propagation facets can be seen as
the reason for this discrepancy. The difference between dwell propagation and
320 initiation facets has previously been pointed out by Pilchak and Williams [23].
The observations made here on initiation features are consistent with this work.

The crystallographic plane of the crack initiating grain did not differ between
dwell and LCF. The grains fracture near the (0002) plane, which has been doc-
umented [24, 22, 20]. The misorientation is reported to be smaller in facets
325 formed by continuous cycling than dwell facets, but the difference in misorien-
tation found here was too small for a clear distinction between the load regimes.
Moreover, this study has shown that facets can be formed that span more than
just one grain. In such circumstances, it is impossible to determine which grain
initiated facet formation. To emphasise the point, the smallest misorientation
330 was found in an LCF facet, but so was the largest.

When the applied stress increased, the fractographic distinction between
dwell and LCF disappeared. At the highest tested LCF stress of 940 MPa, the
fracture initiation behaviour changed into a subsurface initiation typical for the
tested dwell samples. Furthermore, a long needle like structure was seen close
335 to the initiation, thought to be associated with a macrozone. Consequently, as
the stress was increased the fracture became more “dwell-like”. In all observed
facets the average misorientation was smaller than $\sim 15^\circ$, which is the angle
between the $\{10\bar{1}7\}$ hydride habit plane and the (0002). Pilchak et al. [5]
observed a dwell propagation facet that fractured at this misorientation in Ti-
340 8Al-1Mo-1V. Pilchak et al. found small particles on the dwell facets which were
argued to be hydrides. In the present study, the dwell subsurface initiation facet
did not fracture near the hydride habit plane, nor were there particles found.
An influence of hydrogen on initiation of dwell or LCF facetting is therefore
not indicated by the present work. **However, the propagation of dwell facetting**

345 along macrozones has previously been observed by Pilchak et al. [5, 25].

4.2. Spatial orientations

Striated grains showed generally smaller tilts than any faceted feature. Moreover, their normal was tilted relative the rolling direction. This preferential tilt is thought to be a result of the texture. As seen in Figure 2 the majority of
350 the α -Ti grains were orientated with their $\{0002\}$ plane perpendicular to the rolling direction and $\{11\bar{2}0\}$ along the normal direction. These are the ‘hard’-oriented macrozones. In contrast, ‘soft’-oriented macrozones have their basal plane along the rolling direction. Bantounas et al. [26] found that the c -axis of striated grains was tilted towards the loading direction in a range of $55 - 87^\circ$.
355 Thus, the striated regions appear to be associated with soft-oriented macrozones, with their $\{0002\}$ in the RD.

A preferential spatial orientation was also found in the subsurface initiation facets caused by dwell. These were tilted $\sim 30^\circ$ from the transverse direction, with a near- (0002) facet plane ($< 10^\circ$ away). This orientation is inconsistent
360 with the facets being perfectly TD-oriented, *i.e.* from the hard-oriented macrozones. Instead, it suggests that the dwell subsurface initiation facets form from grains in soft macrozones; consistent with the Stroh model these would be at the hard/soft macrozone boundary.

Apart from the fact that dwell in combination with LCF increased the number of initiation sites, the facet spatial orientation did not change. The subsurface initiation behaviour of the dwell facets was found to be the crucial factor. The fact, that the LCF 880 propagation facets seemed to populate one quadrant of the stereographic projection, could indicate a preferential orientation - with facets tilted towards the crack nucleation site.

370 4.3. Crack propagation

Under dwell the crack grew mainly along macrozones. The facet propagation speeds have been reported to be between 2 and 20 μm per cycle and almost

independent of ΔK [5, 13]. In some cases dwell facets propagate up to two orders of magnitude faster than cyclic striations [25]. This means that just after
 375 initiation, the crack propagates along a macrozone and could stretch along the whole sample within a few hundred cycles. The primary dwell faceted region was 1350 μm in length and could have grown in as few as 675 cycles. Because multiple initiation occurred, each growing along a macrozone, the area supporting the applied load reduced quickly until it was below a critical threshold and
 380 fractured.

Crack growth behaviour under variable amplitude loading has been extensively studied for different metals. An overview of sequences and observed material behaviour has been provided by Doré et al. [27]. The consecutive variation in load amplitudes can be described as a series of step-up and step-down changes.
 385 There, a step down change in load was found to cause a reduction in propagation speed, step up changes initially increased the growth rate, which then reduced to the standard rate for that loading. A sequence of single or multiple high stress loads has also been described in the literature as peak overloads [16]. Essentially, an overload is a large step up and step down load sequence. Depending on the overload ratio and ΔK , crack propagation can either accelerate
 390 or retard.

The LCF 880/940 sample (Z) showed crack retardation in both load blocks compared to the baseline tests. Ward-Close et al. [16] argued that the overload initially speeds the crack up because of reduced crack closure near the
 395 tip. The closure is re-established during the overload block, but an increased residual compressive stress field around the crack tip remains. This then causes improved closure at the crack tip, reducing the growth speed. Also discussed in the literature is the possible blunting of the crack by overloads [28]. The initial sharp crack after a 880 MPa cycle could have blunted after the 940 MPa cycles, reducing the stress concentration and hence slowing the crack. However, blunting
 400 was seen to slow crack propagation only at high overloads. An effect from increased surface irregularities contributing to roughness-induced crack closure has also been observed in literature [28]. In the present case this seems unlikely,

since no increased roughness was seen during SEM observations.

405 The crack growth behaviour in the LCF 880/820 sample (Y) was surprising. Crack growth in the low stress 820 MPa blocks were not influenced by the higher loading cycle. However, crack growth acceleration during the 880 MPa blocks was observed relative to the simple 880 MPa LCF test. Following the logic that overloads slow a propagating crack, the 820 MPa blocks should have shown a
410 reduced crack growth rate. This leads to the conclusion that the stress 880 MPa was simply not sufficient to cause enhanced residual compressive stresses at the crack tip. The 0.2% yield stress of the material is 950 MPa, hence there was not enough plastic deformation introduced at 880 MPa to cause a retardation effect. The 880 MPa block showed an increase in crack propagation rate which could
415 be contributed to the initial speed-up, similar to the findings of Ward-Close et al. [16].

The dislocation study carried out on the LCF 880/820 (Y) and LCF/Dwell 880 samples (X) showed that two pyramidal slip systems were activated during striation formation. A change in striation separation could not be correlated
420 with a change of the slip plane separation, but a deflection of the crack path was observed. This step is thought to be caused by a break up in the symmetry of the slip at the crack tip. Dislocation glide in one of the two pyramidal slip bands increased, deflecting the crack path along that plane. This effect was most pronounced in the change from continuous cycling to dwell hold. The two
425 minute dwell allowed for so much slip that a step ten times higher than LCF striations was created. Screw dislocations are known to show time-dependent slip in Ti-6Al-4V, supporting this hypothesis [29, 30].

4.4. Dislocation networks in the as-received material

The dislocation networks found in the as received material are low energy
430 configurations formed during heat treatment [31]. The network is formed by two screw dislocations on different, non-parallel planes. The parent dislocations with different Burgers vector intersect and form a third dislocation which lies on a slip plane that shares a common zone axis with the parent disloca-

tions. The Burgers vector of the newly formed dislocation is the sum of the
435 two initial dislocations, forming a stable junction [32]. The driving force for
junction formation is the reduction in overall dislocation length by the growth
of the common segment. The dislocation network in the as-received material in
Figure 9a is formed by dislocations on the $(1\bar{1}01)$ and $(10\bar{1}1)$ pyramidal planes
and the $(01\bar{1}0)$ prismatic plane. Dislocation analysis confirmed that the Burg-
440 ers vector of the new segment was the sum of the parent dislocations formed
according to the reaction $\frac{1}{3}[11\bar{2}0] + \frac{1}{3}[1\bar{2}10] = \frac{1}{3}[2\bar{1}\bar{1}0]$. The possible formation
mechanism is depicted in Figure 13. Reactions between $\langle c + a \rangle$ screw disloca-
tions have also been seen to form networks [33]. Hexagonal networks of screw
dislocations are low angle twist boundaries [31]. The networks accommodate
445 small twists between grains of similar orientation. Consequently, they can be
found in fracture-initiating grains because these frequently consist of two or
more similarly-orientated grains. Such networks were found in subsurface ini-
tiation facets, which were observed in the EBSD data to consist of more than
one grain. Dislocation networks were indicated by Wang et al. to be involved in
450 fracture initiation [34]. Triple junctions between networks on different planes or
between slip bands were reported at dwell crack initiation. Furthermore, Wang
et al. proposed that cracks could propagate preferably along the dislocation
network due to the stored energy within them. Such networks along the basal
plane might offer a explanation for facetting along the (0002) plane.

455 [Figure 13 about here.]

4.5. Dislocation networks after fatigue

The network in the subsurface facet of the LCF 820 MPa sample (B) that
was found 450 nm below the surface showed little sign of distortion and no sign
of dislocation pile ups. However, below the fracture surface the remains of a
460 dislocation network are seen. This could be an indication that cracks propagate
along networks easily. However, in the as-received material dislocation networks
along the basal plane were not observed. The electro-polished foil taken from

the gauge section of the 820 MPa sample (B) showed that plastic deformation was limited to grains along the crack growth path.

465 Under dwell at 880 MPa dislocation activity increased, as seen in the entangling of the network and pile ups of $\langle c + a \rangle$ screw dislocations. In the long dwell periods dislocations have enough time to unlock from a sessile configuration. Dislocations were seen by Zhang et al. to be transmitted through a network or absorbed by it [35]. In the present study dislocations were observed to pile up
470 at networks and they cross slipped on a plane which was along the net. The fact that only $\langle c + a \rangle$ pile ups were found might be connected to the observation that the network dislocations were missing their $\langle c \rangle$ -component dislocations, although this is speculative.

In the literature it is argued that dislocation networks form during LCF cycling [36] in Ti-6Al-4V at room temperature. Networks similar to the ones found
475 in this work were observed after fifty cycles at 200 ± 775 MPa. However, Kang et al. make no comparison to network structures in the as-received material and do not show if new networks are generated by fatigue. In the present work there was no evidence visible that networks form during continuous cycling. Networks
480 were only seen to disintegrate during dwell or by cycling at stresses near the yield strength. In dwell the networks were seen to entangle. Wang et al. has put forward the idea that dislocation networks can develop during dwell [34]. The fact that Wang also found networks prior to dwell fatigue casts doubt on that theory. In a previous high temperature compression experiment, Wang et al.
485 reported a dislocation network consisting of two $\langle c + a \rangle$ and one $\langle a \rangle$ -type screw dislocation [33]. This experiment was conducted at 1020°C where a dynamic generation of dislocation network by thermal activation would be possible. To prove that the networks are not only formed during heat treatment, in situ experiments would need to be carried out.

490 4.6. Ordering and superdislocations

TEM diffraction patterns have shown the presence of diffuse Ti_3Al superlattice spots prior to testing. In Ti_3Al dislocations are known to split into partials

with an antiphase boundary between them [37]. Under TEM these two superpartials would be visible as a double line. This dissociation is categorised into two types - mode I (wide dissociation type: straight screw dislocation) and mode II (narrow dissociation type: curved screw dislocation) [38]. The visible double lines in the dislocation network and surrounding dislocations seen in Figure 9(c-d) indicate dissociation into superpartials. The dislocations in the network are mode II partials, whilst the wider-separated $\langle 2c + a \rangle$ pyramidal dislocations are mode I superpartials. Examples of networks formed by superpartials in Ti_3Al were observed by Legros et al. [39]. Interactions between Superdislocations have been predicted to cause microcracks [40]. According to Kar'kina a reaction between an $\langle a \rangle$ -type dislocation on a prismatic plane and a $\langle 2c + a \rangle$ -type on a pyramidal plane, as was the case in Figure 9c, could result in fracture along the (0002) plane.

5. Conclusions

Equiaxed Ti-6Al-4V UD plate has been subjected to un-notched low cycle and dwell fatigue, in order to study the effect of block loading regimes on the fatigue cracking behaviour. The following conclusions can be drawn,

1. Underloads cause an increase in crack propagation speed during subsequent reloading, whilst overloads at stresses near the yield strength slow the crack down. Individual dwell cycles during LCF were found to be especially damaging. For the conditions examined here the retardation effect of overloads could only be observed by a smaller average propagation rate per cycle, not by a reduction in the number of striations (crack arrest).
2. Initiation and propagation features could be distinguished by their preferred spatial orientation, flatness or tilt towards the tensile axis. Dwell initiation facets were smoother than LCF facets, were more inclined towards the loading direction and were always found subsurface. During propagation, dwell facets were formed within the macrozones and were

found to be normal to the loading direction. The texture of the material resulted in a preferential spatial orientation of crack initiation and growth features.

- 525 3. Initiation facets were typically formed by more than one grain, fractured on a plane which was $2-13^\circ$ away from the basal plane and was not a hydride habit plane. Subsurface initiation facets were generally not found within 10° of the TD ($\parallel$ LD) orientation, and were on average $30-35^\circ$ from the TD. This is in contrast to the Stroh model, which postulates an (0002) 530 facet normal to loading, in the TD orientation, initiated by shear in an adjacent grain that is well oriented for slip.
4. Dislocation networks found in α -Ti are twist boundaries between similarly oriented grains that form at high temperature and were found in initiation facets. Dislocation networks can be obstacles to dislocation motion and 535 promote cross slip. These dislocation networks disintegrated under dwell loading or high stresses.
5. Changes in load caused deviation of the crack growth direction in striated grains, which contributed to a loss of symmetry in the slip at the crack tip. A correlation between striation separation and slip band separation 540 could not be observed.

Acknowledgements

The material used in this study was supplied by Timet during EPSRC project EP/E044700/1. DD was supported by EPSRC grants EP/H004882/1 and EP/K034332/1, and POT by a Rolls Royce plc - EPSRC DTA CASE 545 studentship. We thank Dr Catrin Davies for the use of the facilities in the Department of Mechanical Engineering at Imperial College for some of the testing.

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	Load	N_{LCF820}	N_{LCF880}	N_{LCF940}	$N_{Dwell880}$	Σ	# Initiation Sites	
							Surface	Subsurface
A	LCF 880 MPa	0	21843	0	0	21843	1*	-
B	LCF 820 MPa	278908	0	0	0	278908	1*	1
C	LCF 940 MPa	0	0	9613	0	9613	1	1*
D	Dwell 880 MPa	0	0	0	9310	9310	-	>9*
X	10A/1D	0	7349	0	733	8072	-	>24*
Y	10A/40B	50596	12650	0	0	63246	1*	-
Z	100A/10C	0	21642	2160	0	23802	1*	2

Table 2: Comparison between TEM and QTF/EBSD measurements of the misorientation between the (0002) pole and plane of fracture.

Sample	Surface/ Subsurface	Angle ($^{\circ}$)	
		TEM	QTF/EBSD
LCF 880 (A)	Surface	8.9	7.0, 12.2
LCF 820 (B)	Surface	10.6	2.4, 10.6
	Subsurface	4.2	1.8, 4.6, 7.5
LCF 940 (C)	Surface	10.8	13.4
	Subsurface	3.5	3.0
Dwell 880 (D)	Subsurface	-	6.3
LCF/Dwell 880 (X)	Subsurface	-	8.5, 12.3
LCF 880/820 (Y)	Surface	-	4.3, 7.5
LCF 880/940 (Z)	Surface	-	4.9