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Concurrent operation of ⟨c + a⟩ slip and twinning under cyclic loading of Ti-6Al-4V

C. Lavogiez, S. Hémery *, P. Villechaise

Institut Pprime, CNRS-ENSMA, Université de Poitiers, UPR CNRS 3346, Physics and Mechanics of Materials Departement, ENSMA – Téléport 2, 1 avenue Clément Ader, BP 40109, 86961 Futuroscope Chasseneuil Cedex, France

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Deformation processes operating in hard/soft grain pairs were characterized in Ti-6Al-4V specimens with a bimodal microstructure submitted to fatigue and dwell fatigue loadings. Concurrent operation of 1st order ⟨c + a⟩ pyramidal slip and slip-stimulated $\{10\bar{1}2\}\langle10\bar{1}1\rangle$ deformation twinning was observed in hard grains, resulting from multi-scale strain incompatibilities. The former constitutes a direct experimental evidence of stress redistribution under cyclic loadings while the latter, which is triggered by local stress concentrations due to dislocations pile ups, is detected for the first time. Insights into the crack initiation processes are then discussed.

Titanium alloys are widely employed in the aerospace industry owing to their superior corrosion resistance and mechanical properties combined with a relative low density. For gas turbine applications, fatigue performance is fundamental as in-service conditions induce cyclic loadings. The introduction of a hold time at peak stress leads to a loading that is more representative of flight operation. For a great number of titanium alloys, this leads to a substantial life debit, known as the “dwell effect” [1]. Both experimental and modeling approaches have been applied to identify the origin of this dwell effect, in order to propose microstructural evolution to limit or to suppress its detrimental consequences [2–5].

Pioneering works revealed the presence of basal or near-basal facets at crack initiation sites in grains poorly oriented for basal or prismatic slip (denoted in the following as hard grains) due to the close alignment of its c-axis with the loading direction [6]. Evans and Bache suggested a modified version of Stroh's model in order to account for this experimental observation [2]. According to the authors, crack initiation is a consequence of basal or prismatic slip operating in an adjacent grain (denoted in the following as soft grain). Dislocations piled-up at the boundary with the hard grain generate the required stress state to induce facet formation. A key contribution comes from soft regions off-loading stress on to hard regions. Indeed, the difficult ⟨c⟩ axis deformation, implied by a high critical resolved shear stress (CRSS) for ⟨c + a⟩ pyramidal slip [7] and twinning inhibition due to high aluminum content [8], leads to important strain heterogeneities [3]. As a result, load shedding occurs in the regions with low plastic strain accumulation (i.e the hard grain). The stress magnitude in the hard grain may exceed then the remote applied stress and triggers crack initiation [4, 9].

Crystal plasticity based simulations have provided novel insights into the stress redistribution and its modeling. However, they often fail at capturing the real local processes resulting from strain localization such as stress concentrations induced by dislocations pile-ups. Experimental characterization at the relevant scale being still scarce [10], the deformation processes operating in soft/hard grain pairs before crack initiation deserves a dedicated study. This is the aim of the present article. Since striking similarities between crystallographic orientations at crack initiation sites were observed under fatigue and dwell-fatigue loadings [5], deformation processes under fatigue loading were presently investigated as well.

The material studied is a Ti-6Al-4V alloy with a bimodal microstructure. The surface fraction of the primary nodules is 36% and the mean diameter is 13 µm. Two flat dog bone specimens with 10 mm long, 2 mm wide and 1 mm thick gage length were tested. A solution of nine parts of colloidal silica, with 0.04 µm particle size, and one part of 30% H_2O_2 was used as the finish polishing on one face of the specimens.

In order to highlight behavior differences between fatigue and dwell-fatigue loadings, the loading conditions were chosen to induce a significant dwell fatigue life debit. The load waveforms employed are shown in Fig. 1a. The maximum peak stress was set to 908 MPa (corresponding to 0.94 x $\sigma_{0.2}$) and the R ratio to 0.1 [11, 12]. Loading and unloading were achieved in 1 s. The hold time at minimum stress was 1 s. Fatigue and dwell-fatigue conditions involves 1 s and 120 s load hold respectively. The 0.2% proof stress ($\sigma_{0.2}$) had been previously

* Corresponding author.
E-mail address: samuel.hemery@ensma.fr (S. Hémery).

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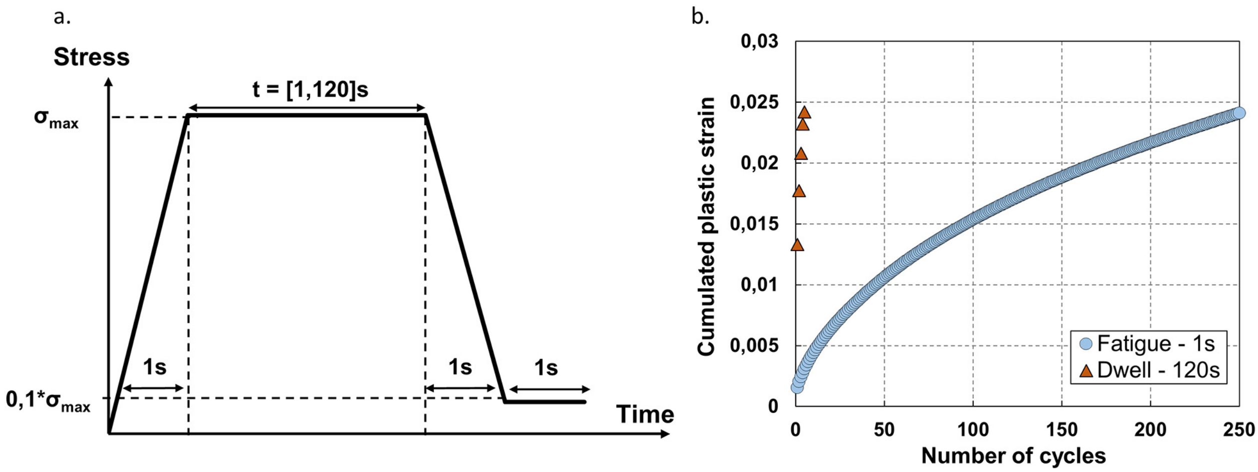

Fig. 1. a. Cyclic load – time waveforms applied for fatigue and dwell-fatigue tests, b. cumulated plastic strain against the number of cycles before test interruption showing a reduced number of cycles needed to reach 2.5% plastic strain for the dwell-fatigue test.

estimated using a tensile test with an equivalent strain rate. An Instron 1362 universal testing machine was employed to test the specimens up to a plastic strain of about 2.5%. The cumulated plastic strain against the number of cycles is presented in Fig. 1b. 250 fatigue cycles were necessary to reach the targeted plastic strain versus <5 dwell-fatigue cycles. Following the mechanical testing, the electron backscattered diffraction (EBSD) technique was employed to map the crystallographic orientation over large regions in order to identify regions of interest containing hard-soft grain pairs. EBSD characterization with a step of 70 nm was performed using a JEOL 7000F scanning electron microscope (SEM) equipped with an EBSD setup provided by EDAX. Local crystallographic orientation data also enabled the identification of activated slip systems through slip trace analysis and apparent Schmid factor calculation. The reader is referred to prior studies involving similar procedures for further details about mechanical testing, slip trace analysis and apparent Schmid factor calculation [12, 13]. For clarity purposes, only some typical cases are detailed in the present article.

An SEM micrograph of a soft – hard grain pair identified in the dwell-fatigue tested specimen is shown in Fig. 2a. Coarse slip traces in grain A indicate extensive plastic deformation through intense and highly localized slip activity. Slip trace analysis reveals the occurrence of prismatic slip with an apparent Schmid factor of 0.49. Interestingly, this is the slip system with the highest apparent Schmid factor among basal, prismatic and pyramidal slip systems. Slip bands seem blocked at the interface with gain B, which can be considered as hard according to both the low misorientation between its c-axis and the loading direction (9°) and the low apparent Schmid factor for basal and prismatic slip. Nevertheless, thin slip traces matching with a ⟨c + a⟩ 1st order pyramidal slip

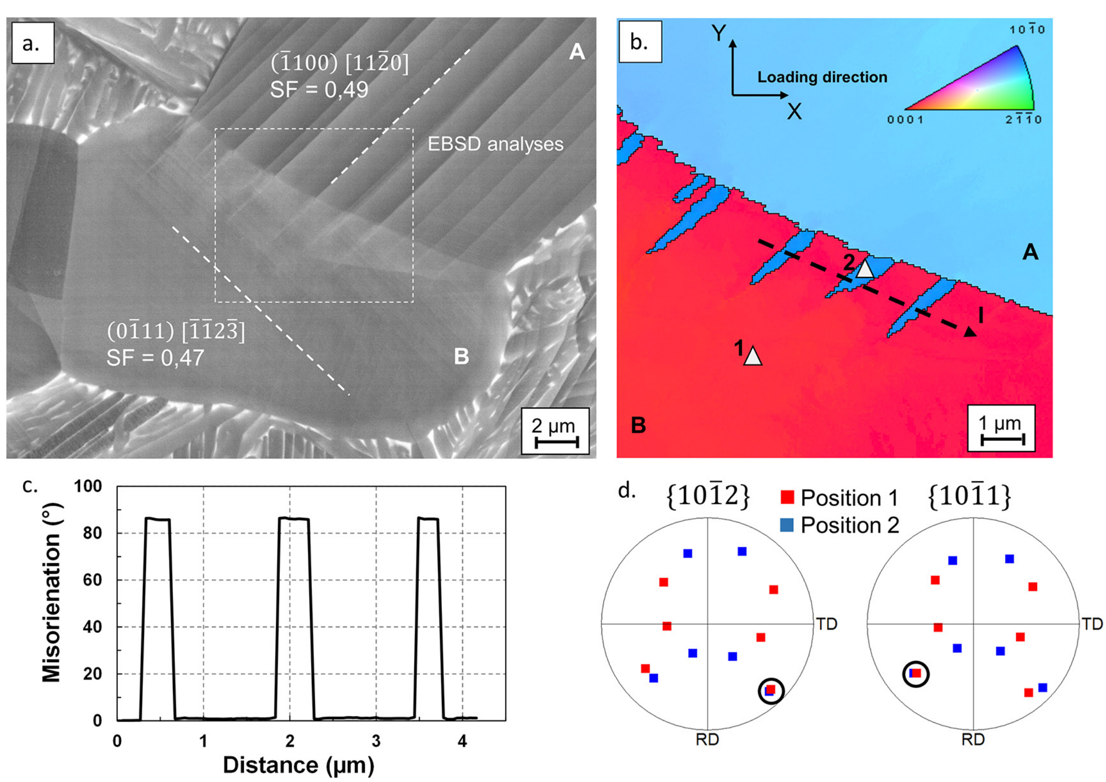

Fig. 2. a. SEM micrograph of a soft/hard grain pair of the dwell-fatigue tested specimen, b. the corresponding inverse pole figure map along the loading direction, c. the misorientation profile along path I and d. $\{10\bar{1}2\}$ and $\langle 10\bar{1}1\rangle$ pole figures at positions 1 and 2.