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Dislocation arrangements within slip bands during fatigue cracking

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A B S T R A C T

Microstructures and dislocations associated with the slip band (SB) fatigue cracks were investigated in Cu single crystal and bicrystal. It is found that there are high strain localizations with abundant dislocation cells around the cracked SBs. Due to the high local strain and slip irreversibility, the lattice rotates severely leading to large misorientation and hence low- or high- angle grain boundaries could form between the neighboring cells. Moreover, the dislocation walls can serve as preferential sites for nucleation and propagation of fatigue cracks with high density of dislocations, vacancies and high strain incompatibility.

1. Introduction

It is well-known that during the cyclic deformation, slip bands (SBs) would form with abundant extrusions or intrusions on the specimen surface along with the dislocation slip. The SBs are widespread in all kinds of crystalline materials and it has been found that the SBs are connected with the evolution of dislocation substructures with extensive and long-range research. To be specific, regular dislocation patterns such as dislocation veins or patches, ladders, labyrinth and cells can form within the SBs [1,2]. It occurs easily in face-centered cubic (fcc) single crystals, bicrystals and polycrystals, such as Cu, Ni, Ag and Cu alloys [3]. More importantly, as cycling continued, the surface SBs are becoming more and more and the fatigue cracks can nucleate along the SBs or the interface between the SBs and adjacent matrix [4]. Due to the annihilation of edge dislocations, vacancy can form predominantly in the SBs [5]. This process is the main cause of slip irreversibility and the concomitant surface roughening. And then the resultant local stress concentrations can facilitate the SB fatigue cracking [5]. It is also believed that a large number of vacancies can form and then aggregate into a microcrack within the SBs [6]. Besides, there is also internal stress existing along the SBs-matrix interface due to the deposition of edge dislocations which is responsible for the crack initiation along the SBs [5,6]. That is, both the internal microstructure and the surface morphology promote the fatigue cracking of SBs.

During the cyclic deformation, different dislocation patterns would form in materials of different fatigue cycles or in different materials. For example, the dislocation patterns could evolve from veins to ladders, labyrinth or cells gradually in the cyclically deformed Cu single crystals [1–3,7]. Yet, there are no such regular dislocation patterns but just dislocation arrays or tangling in some fatigued Cu alloys with low stacking fault energy (SFE) [8,9]. It is significant to correlate the microstructure especially the dislocation arrangements with the fatigue cracking at SBs so as to improve the fatigue cracking resistance. Due to the regularity of dislocation patterns, intensive attention has been paid on the persistent SB fatigue cracking of Cu. Transmission electron microscope (TEM) observation has provided microscopic evidence for the initiation of persistent SB fatigue cracks in Cu bulk specimens and showed that microcracks were initiated and propagated along the ladder-like structures [10]. Recently, the pinpoint analysis on the persistent SBs with the aid of focus ion beam (FIB) milling also confirms that fatigue cracks initiate from the intrusions of persistent SBs having ladder-like dislocation structure [11]. So far, great efforts have been made to study the fatigue crack nucleation from the surface. In fact, with further cyclic deformation, SB cracks would grow and cell structures could form around the SB cracks [10]. Besides, the crack propagation is also a crucial concern during the fatigue process which often consumes a large part of fatigue life. The present study aims to detect the early growth path of the SB fatigue cracks and the corresponding microstructure evolution (especially the dislocation patterns) which can help to further understand the cracking mechanism as well as to provide some references to improve the fatigue performance of material.

2. Materials and Experiments

Cu single crystal and Cu bicrystal with an incoherent twin boundary (ITB) grown from oxygen-free high-conductivity copper of 99.999% purity by Bridgman method were employed. Their crystallographic orientations were determined by electron backscatter diffraction

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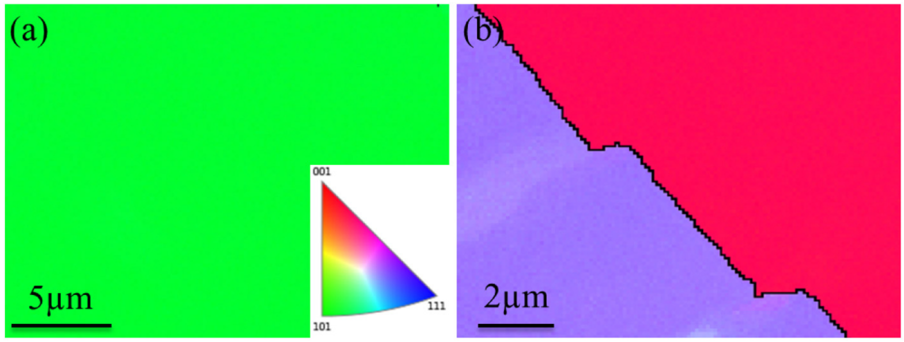

Fig. 1. EBSD detections of Cu single crystal and bicrystal before cyclic deformation. (a) The GB and IPF map of the single crystal, (b) the GB and IPF map of the bicrystal with an ITB. The color of the IPF represents the in-plane orientation parallel to the loading axis of the specimens.

(EBSD) method. The inverse pole figure (IPF) and grain boundary (GB) maps in Fig. 1 show that all the grains of the samples are single-slip oriented and the ITB is inclined at ~45° to the loading direction. Then they were cyclically deformed under symmetrical push-pull loads at increased shear stress amplitude. The shear stress amplitudes increased stepwise from 10 MPa to about 32 MPa and were finally maintained at about 32 MPa. A triangular wave with a frequency of 1 Hz was used the same as our previous work [12–14]. The specimens were deformed until the fatigue cracks reached hundreds of micrometers in length which can be detected by the Zeiss Supra 55 Sapphire scanning electron microscope (SEM). Subsequently, the specimens were examined by transmission electron microscope (TEM), electron channeling contrast (ECC) and EBSD techniques within the SEM to investigate the dislocation arrangement in the SBs.

3. Results and Discussion

3.1. Typical Morphologies and Dislocation Arrangements of SB Fatigue Cracking

The deformation and fatigue cracking morphologies of both the single crystal and bicrystal are shown in Fig. 2. It can be seen from Fig. 2(a) that though some secondary SBs operate, the fatigue cracks nucleate and propagate along the primary SBs preferentially as found previously [4–6,10,11]. In the bicrystal, the SBs can pass through the ITB continuously since they have common slip plane and slip vector in both component grains. The fatigue cracks also nucleate and propagate along the SBs firstly and even some SB cracks can penetrate the ITB as shown in Fig. 2(b). It is noteworthy that the fatigue cracks can nucleate along the SBs or the interface between the SBs and the matrix. In fact, the SB fatigue cracks are common in polycrystals and the typical SB fatigue cracking in single crystal and bicrystal are employed here for simplicity.

The dislocation arrangements of the SBs in Cu bicrystal were detected by TEM and displayed in Figs. 3(a, b). Due to the inhomogeneous deformation, dislocation patches, ladders and cells all can form in one sample. The typical ladder and cell structures are shown in Fig. 3(a) and (b), respectively. Generally, the dislocation ladders demonstrate lower strain than dislocation cells [1–3,7]. In comparison with TEM method, ECC detection is more suitable for observing the dislocation patterns near the SB fatigue cracks in a larger scale as shown in Fig. 3(c, d). The surface relief was ground off and the samples were electro-polished before the ECC observation. ECC imaging is produced by using the conventional image scanning with a back-scattered electron detector in the SEM. The working distance is about 8 mm and the acceleration voltage is 20 kV using an aperture of 60 mm. The SB fatigue cracks are visible and exhibited as white band in Fig. 3(c, d). Although the edge of cracks can be etched off easily during the electropolishing, the general alignment of the fatigue cracks can still be easily distinguished to be along the SBs in Cu single crystal and bicrystal. Moreover, the morphology of narrow and small fatigue cracks in the bicrystal in Fig. 3(d) is little affected by electropolishing. In the Cu single crystal and bicrystal, all SB fatigue cracks are surrounded by dislocation cells. The dislocation cells are of varying sizes, from a few hundreds of nanometers to a few micrometers. Qualitatively, the cell sizes in areas near the crack are smaller than that far away. It is believed empirically that smaller size of dislocation cells comes from higher stress or strain [15]. That is, the local strain should be higher in the vicinity of the SB fatigue cracks than the areas far away.

3.2. EBSD Detection on the Strain Concentration near the Cracked SBs

The microstructure near the grown SB fatigue cracks were further analyzed with the help of EBSD method as illustrated in Fig. 4. For EBSD measurement, an accelerate voltage of 20 kV, a lens of 60 mm and a working distance of ~13 mm were used. The step size is 50 nm with the indexed rate higher than 90%. The IPF and GB maps along with the corresponding band contrast (BC) maps near SB fatigue cracks in the single crystal (Fig. 4(a,b)) and bicrystal (Fig. 4(d, e)) reveal that there is significant difference between the crystallographic orientations of dislocation cells. As indicated by the orientation triangles in detail, the lattice rotates without regular pattern during the cyclic deformation. It

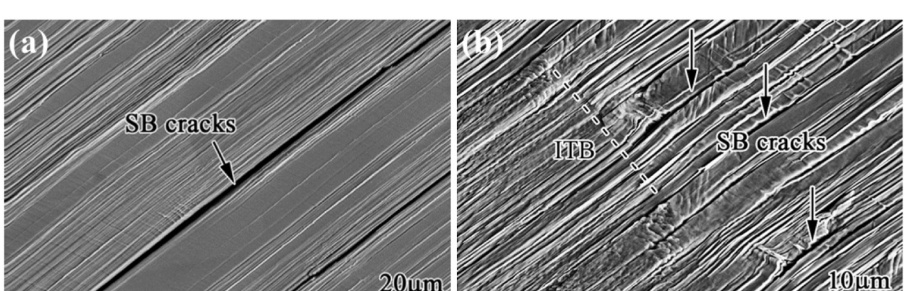

Fig. 2. Typical SB fatigue cracking in (a) single crystal and (b) bicrystal.