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Unified factor controlling the dislocation evolution of fatigued face-centered cubic crystals

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ABSTRACT

By analyzing the formation mechanisms of different types of dislocation patterns, one unified factor is developed to account for effects of slip mode and further explore the basic law of dislocation evolution in fatigued face-centered cubic (fcc) crystals. First of all, in the formation of persistent slip band (PSB) ladders, geometrically necessary dislocations (GNDs) accommodate the elastic/plastic strain gradients between the hard walls and soft channels and provide the long-range internal stresses required for the simultaneous compatible deformation of soft and hard regions. In typical wavy-slip materials, advanced dislocation patterns include the wall, cell and labyrinth structures. The formation of deformation band (DB) walls may be derived from the accumulation of GNDs. The appearance of labyrinth and cell structures should be related to the activation of critical and coplanar secondary slip systems, respectively. In typical planar-slip materials, the dislocation structures consist of the dipole arrays and stacking fault (SF) bands. The constant compression of the split distance between partials will lead to the conversion of the closely spaced dipole array to the SF bands, which indicates that both wavy-slip and planar-slip materials follow one unified evolution factor. This factor, labelled α in this work, may be described as the ratio of the annihilation distance of screw dislocations to the split distance between partials, which can characterize the slip mode. The higher the α value is, the easier will be the appearance of PSBs and various advanced dislocation patterns. With decreasing the α value, dislocation evolution gradually changes from 3D patterns to 2D structures.

1. Introduction

Basic mechanisms of fatigue have been investigated for decades and most attention has been paid to wavy-slip materials in face-centered cubic (fcc) crystals since the discovery of persistent slip bands (PSBs) [1]. In wavy-slip materials the dislocations can easily change their slip plane by cross slip, which is essential for the formation of 3D composite structures [2–4]. In the classical textbook "Fatigue of Materials", Suresh [5] systematically summarized the research results of different kinds of fatigued fcc crystals [6–16]. At room temperature the saturation shear stresses in the prominent plateau of the cyclic stress-strain (CSS) curves for single-slip oriented Cu and Ni single crystals are 27.5 MPa and 50 MPa, respectively, while the saturation plateau stress of single-slip oriented Ag single crystal is only ~20 MPa [17]. It should be pointed out that the ratio of the saturation shear stress to the shear modulus is remarkably close for all the three materials, which suggests that the evolution of their dislocation structures follows a similar process under cyclic loads.

First of all, PSB-ladder structure corresponding to the plateau behavior originates from the metastable multipolar vein structure. The early experimental evidence for the metastability can be traced back to the research from Neumann et al. [18,19] and Basinski et al. [20]. The formation of the edge multipole veins can be attributed to the effective elimination of screw dislocations by cross slip. Thus a so-called wavy cyclic slip mode, characterized by easy cross slip, favors the formation of the metastable vein structure and the transformation into the PSBs. In the 1960s, Feltner and Laird [2,3] confirmed that fcc metals with different slip character showed distinct sensitivities to the initial condition by comparing Cu and Cu–Al alloys. Moreover, they displayed the respective dislocation patterns for different fatigue lives in a map and suggested a simple

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connection between their formation and the stacking fault energy (SFE) [4]. Based thereupon, Lukáš and Klesnil [21] later explicitly described the strong dependence of the types of dislocation substructures on the SFE and the plastic strain amplitude by comparing Cu and Cu–Zn alloys.

Since the work of Feltner and Laird [4] as well as Lukáš and Klesnil [21], it had become customary to characterize the cyclic slip mode by the SFE. However, Blochwitz et al. [22] found that alloys of high yield strength can exhibit planar slip in spite of a relatively high SFE, such as Ni–Co and Ni–Fe alloys. Furthermore, Karnthaler and Schügerl [23], Clement [24] and Steffen et al. [25] pointed out that in such alloys short-range order (SRO) plays an even more important role in determining the slip mode. Thus, in order to characterize the slip mode, Mughrabi and Wang [26] proposed an annihilation distance y_s of screw dislocations, depending on different metallurgical parameters, to represent a more suitable material parameter. In 2001, Hong [27] summarized the relationships between the CSS response and slip mode of Cu–Al and Cu–Zn alloys with different Al and Zn contents, respectively. He suggested that the slip planarity was promoted not only by low SFE, but also by the increase in shear modulus, atomic size misfit and solution content. Accordingly, he developed a model for discriminating wavy- and planar-slip behaviors quantitatively. According to plenty of research results over nearly half of a century [6,10,11,28–38], Wang [39] further concluded that although on many occasions the value of SFE can be employed as the criterion for distinguishing wavy slip from planar slip, one cannot always obtain a satisfactory result if such a criterion for the transition is employed alone.

Therefore, despite plenty of experimental and theoretical results, it is still quite necessary to consider various fcc crystals as a whole and to find a new, more generalized factor controlling the basic mechanisms for dislocation evolution in fatigued fcc crystals with different slip modes. Based on the above analysis, Cu, Ag and Cu–Al alloys representing different slip modes are selected as model materials to carry out the low-cycle fatigue experiments. Furthermore this study will mainly deal with two topics. One is to explore the formation mechanisms of all different types of dislocation patterns. The other is to clearly outline our present understanding on the basic laws for dislocation evolution in various fatigued fcc crystals based on one unified evolution factor.

2. Experimental procedures

Cu and Ag single crystals were firstly grown from electrolytic Cu of 99.999% purity and Ag of 99.999% purity by the Bridgman method in a horizontal furnace. Secondly, Cu-5at% Al, -8 at% Al and −16 at% Al non-textured polycrystalline alloys with average grain size of 1–2 mm were prepared from electrolytic Cu (99.999 pct) and Al (99.999 pct) by melting in a vacuum of 10^{-4} mm Hg. Later, fatigue specimens with a gauge section of 7 mm × 5 mm × 16 mm and total length of 54 mm were made using an electro-spark cutting machine.

The crystal orientation was determined by the electron backscattering diffraction (EBSD) technique with an accuracy within ±2° in a Cambridge S360 scanning electron microscope (SEM), including $[\bar{1}39]$ Cu single crystal and $[\bar{2}39]$ Ag single crystal. Their Schmid factors are respectively 0.493 and 0.478 and the corresponding ratios of the second highest to the highest Schmid factor are 0.947 and 0.92, respectively. Cyclic deformation was performed under symmetric push-pull loading at room temperature in air using a Shimadzu servo-hydraulic testing machine. A triangular waveform signal with a frequency of 0.5 Hz was used for the constant plastic strain amplitude control. Plastic strain resolution and control precision were better than 10^{-5}. The peak loads in tension and compression and the hysteresis loops were recorded by computer automatically until cyclic saturation.

After the fatigue tests, all the fatigued specimens were electropolished carefully and dislocation configurations were subsequently observed by the electron channeling contrast (ECC) technique in a Cambridge S360 SEM. The principle for the electron channeling phenomenon has been described in detail by Zauter et al. [40]. Stickler et al. [41] attempted for the first time to apply the electron channeling effect to the study of deformation behavior of materials. Such ECC images are similar in appearance to transmission electron micrographs, albeit with a lower image resolution [42]. In this paper, the $(1\bar{2}1)$ plane was chosen as the observation plane of the dislocation patterns in $[\bar{1}39]$ Cu single crystal. More detailed dislocation structures in fatigued Ag and Cu–Al specimens, especially the interaction between PSB walls and stacking faults (SFs), were investigated with a Tecnai G2 F30 transmission electron microscope operated at 300 kV. The transmission electron microscopy (TEM) thin foils were first sliced from the gauge part of the fatigued specimens by spark-cutting, then mechanically thinned down to a thickness of about 100 μm and finally polishing by a conventional twin-jet method.

3. Results and discussion

3.1. Geometrically necessary dislocations (GNDs) and PSB ladders formation

Table 1 lists in detail the fatigue testing conditions and types of dislocation patterns for cyclic saturation in $[\bar{1}39]$ Cu single crystal, $[\bar{2}39]$ Ag single crystal and Cu–Al alloys, where γ_{pl} is the constant plastic strain amplitude, cyclic No. denotes the cyclic number at cyclic saturation. It can be found from Fig. 1 that the PSB ladders are the most typical dislocation patterns in fatigued $[\bar{1}39]$ Cu single crystal. Each PSB is composed of a periodic array of PSB ladders and matrix channels between the ladders. The parallel rungs from the $(1\bar{2}1)$ plane occupy about 10% of the PSBs by volume and consist of dense arrays of edge dislocation dipole [43]. The dislocation densities in the rungs are two to three orders of magnitude greater than those in the in-between channels containing screw dislocations.

The dislocation heterogeneous distribution gives rise to the formation of hard wall and soft channel. To accommodate the elastic/plastic strain gradients between the soft and the hard regions, some local excess dislocations are induced by and stored in the heterogeneous dislocation patterns, which is referred to as “geometrically necessary” dislocations (GNDs). Nabarro likened the GNDs as a chicken and egg problem [44], but no matter whether the geometry, e.g., sharp interfaces or lattice curvatures, make the dislocations necessary or excess dislocations cause the geometry, Ashby [45] referred to the GNDs on a microstructural scale. As illustrated in Fig. 2(a), The GNDs at the interfaces between the walls and the channels can act as sources of internal stresses, leading to a redistribution of the local stresses and thereby ensuring compatible plastic deformation. As shown in Fig. 2(b), the local flow stresses of PSB walls and channels are τ_w and τ_c, respectively, which are assumed to be related to the corresponding dislocation densities ρ_w and ρ_c, by Taylor-type relationships as follows:

$$\tau_w = kG\mathbf{b}\rho_w^{1/2} \tag{1}$$

And

$$\tau_c = kG\mathbf{b}\rho_c^{1/2} \tag{2}$$

where G denotes the shear modulus, $\mathbf{b}$ is the modulus of the Burgers vector and k a constant equal to about 0.3. In Ashby's model [45], a fundamental relationship was derived for dislocation glide in