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Cyclic deformation and dynamic compressive properties of copper bicrystals

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Abstract

Cyclic deformation and dynamic compressive tests of three copper bicrystals were carried out on a Shimadzu servo-hydraulic testing machine and a split Hopkinson pressure bar (SHPB) apparatus respectively. The post-deformation dislocation structures, grain boundary (GB) serrations and adiabatic shear bands (ASBs) were examined using electron channeling contrast (ECC) imaging in a scanning electron microscope (SEM). After cyclic straining the secondary slip bands were activated near the GB forming a GB affected zone (GBAZ). Microstructures beneath it are dislocation labyrinth or irregular persistent slip bands (PSBs). The saturation stress at GBAZ was calculated to be higher than that in the grain interior. In dynamic compression, the formation of ASBs, was found to be promoted by the GB. Dynamic stress–strain curves were compared with each other for single crystals, bicrystals and polycrystals. Microstructures were also characterized with orientation imaging microscopy (OIM) by electron backscattering diffraction (EBSD) technique. It was found that relatively larger lattice rotations occurred across the ASBs than in the other regions, which can be deduced by the localization of simple shear along ASBs. No recrystallization was found in the ASBs or along GBs in the present circumstance.

Keywords: Copper bicrystal; Grain boundary; Cyclic straining; EBSD; Dynamic deformation; Lattice rotation; Adiabatic shear bands

1. Introduction

The role of grain boundaries in plastic deformation has long been an interesting subject and regained increased attention [1–7]. It is well known that with decreasing grain size, the volume fraction of grain boundary (GB) increases significantly. Therefore, GB becomes a controlling factor in the plastic deformation of fine-grained polycrystals. To clarify the GB effects, different kinds of bicrystals, such as aluminum [1,6,8,9], Fe-alloys [4,10,11] and copper [12–15], have been widely employed in the cyclic straining or monotonic loading at low strain rates. Experimental results reveal that large angle random GBs become the preferential sites to initiate fatigue crack during cyclic straining, independent of the interaction angle between the GB plane and the stress axis due to the piling-up of dislocations [15] and there was a GB affected zone (GBAZ) in the vicinity of a GB [6,8,13–15]. GB influences the deformation mode and the mechanical properties of polycrystalline materials to a great extent.

On the other hand, engineering materials are often subjected to high strain rate loadings such as high-speed machining, ballistic impact, metal punching and forging. At high strain rates, materials often exhibit deformation instability characterized by the formation of adiabatic shear bands (ASBs) which have different deformation and fracture behaviors in comparison with those occurred under quasi-static or low strain rate loading. Profuse investigations of high strain rate responses of materials are concentrated on polycrystalline materials [16–24]. The experiments indicated that high strain rate deformation instability and the formation of ASBs seem to be strongly dependent on the sample geometry and mechanical properties of a material rather than its local crystallography [18]. Recent studies even found that the ASBs could not be initialized without material inhomogeneity or geometry inhomogeneity [21]. However, to some extent GB can be regarded as a kind of material inhomogeneity, its effect on the high strain rate deformation is not clearly known yet. There may be two reasons. One is that the mechanical effects or the geometry inhomogeneity are severer than the GB's effect.

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For example, in the hat-shaped specimen method [22], the specimens were machined so that intense pure shear stress can be imposed on a narrow region. During dynamic process, ASBs are forced to generate within this region. In this circumstance, GB's effect is 'covered up' by the geometrical effect. Another reason may be that the width of ASBs (about 2–200 μm) often covers the conventional grain size range (5–50 μm). It is hardly for one single GB to manifest its contribution on the dynamic deformation in the polycrystalline materials. Therefore, the contributions of a single GB on the dynamic compression are still not clear.

Furthermore, preferential dynamic recrystallization (DRX) nucleation at the grain boundary on Cu–Si bicrystals at 1023 K was observed at strain rates from 4×10^{-5} to $4 \times 10^{-3}\ s^{-1}$ [25]. DRX is also found in the ASBs in the ballistic impact at the strain rate above 10^5 [21–23]. However, investigations of DRX at the strain rate about 10^3–$10^4\ s^{-1}$ at room temperature are still rare. In the present paper, copper bicrystals with random GB parallel to the loading axis were tested in the strain rate about $4 \times 10^3\ s^{-1}$. The non-existence of DRX, GB serration, dislocation rearrangement and GB's effect on ASB formation along with bicrystals' cyclic deformation properties are reported.

2. Experiments

Bicrystal plates with the size of 10 mm × 55 mm × 260 mm were grown from OFHC copper of 99.999% purity by the Bridgman method. Specimens with the dimension of 7.5 mm × 5 mm × 60 mm were cut from the plates. Three different bicrystals B1, B2 and B3 were prepared with the gauge section of 5 mm × 5 mm × 20 mm. The specimens were carefully electro-discharge machined to make sure that the GB was in the center and parallel to the loading axis. Prior to the fatigue test, the specimens were mechanically and electrolytically polished on all surfaces to ensure that a strain-free and smooth surface layer was obtained.

Push–pull fatigue tests were conducted at room temperature in air using a Shimadzu servo-hydraulic testing machine with a constant axial plastic strain amplitude of 1.5×10^{-3}. A triangular waveform with a frequency of 0.2 Hz was employed and the tests were terminated after saturation at about 10^4 cycles. Then the surface slip morphology was observed. Afterwards, those surface slip traces were removed and repolished to examine the dislocation patterns, especially near the GB, by the scanning electron microscope (SEM)-electron channeling contrast (ECC) technique [26].

After fatigue the gauge section of the specimens was cut into pieces with a dimension of 5 mm × 5 mm × 4 mm then repolished and used for dynamic tests. Why we use fatigued specimens for dynamic tests, the reasons are as follows: (1) during pre-cyclic straining the dislocation patterns were well developed, which can serve as markers to indicate the deformation process during dynamic loading. (2) The ASBs can occur rather easily in the fatigued single crystals compared with the counterparts without pre-cyclic straining [26]. (3) In practice, a metal component may undergo intense cyclic deformation before impact. Therefore, it is necessary to study the dynamic behavior of fatigued specimens.

Those specimens with pre-cyclic straining were named with prefix 'CS'. According to the finial impact axial strain, specimens of this group, say B1, were named as CS-B1-1, CS-B1-2 and CS-B1-3, with the final strains of −0.138, −0.178 and −0.283, respectively. For comparison, another group of bicrystal B1 with the dimension of 5 mm × 5 mm × 4 mm was also prepared. They did not receive pre-cyclic straining treatment before dynamic tests. They were named as UC-B1-1, UC-B1-2 and UC-B1-3 with the final strains of −0.135, −0.197 and −0.215, respectively.

Dynamic compressive tests were conducted on a split Hopkinson pressure bar (SHPB) apparatus. The specimens were assembled between the incident bar and transmitter bar. The strain rate was adjusted to be about 4×10^3 for all the specimens. Before tests, thick steel tubes of different lengths were mounted outside the specimens. Therefore, the incident bar could be immediately stopped after it contacted the steel tube. In the present tests, the total engineering strains in the range of 10–30% could be obtained.

After dynamic tests, all the specimens were examined in a Cambridge S-360 SEM with ECC technique. The orientation imaging microscopy (OIM) by electron backscattering diffraction (EBSD) technique was conducted with a step of 0.5 μm in a field emission scanning electron microscope (FESEM) of LEO Supra 35.

3. Results and discussion

3.1. Bicrystals' orientations

The axial orientations of the three bicrystals were determined by EBSD technique, as listed in Table 1. The axial orientations of the two component grains of bicrystal B1 are $[\bar{2}\,5\,8]$ and $[\bar{1}\,15\,20]$, respectively, with a misorientation of 10.8°. For bicrystals B2 and B3, the misorientations of the two component grains are 8.5° and 10.2°, respectively. Their orientations are also shown in the standard stereographic triangle (Fig. 1). For bicrystal B1 the component grain G1 orients typically single-slip and is close to $[\bar{1}\,3\,5]$ (less than 2.5°), and grain G2 is close to [0 3 4] double slip orientation (also less than 2.5°) that is at [0 0 1]–[0 1 1] side. For bicrystal B2, both grains G1 and G2 orient for single-slip. For bicrystal B3, the component grain G2 is rather close to the $[\bar{1}\,1\,7]$ double slip orientation (about 8°). As can be seen above, for all three bicrystals, their two component grains have almost same misorientations (8–10°) but orient differently. Their saturation stresses and initial harden rates vary a great deal as presented in the following section.

3.2. Cyclic hardening and saturation behavior

Fig. 2 shows the cyclic hardening and saturation curves of the three bicrystals. All the bicrystals display a typical rapid cyclic hardening and then cyclic saturation behavior. Their axial saturation stresses differ from 62 to 72 MPa (Fig. 2(a)). This