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MICROSTRUCTURE AND MECHANICAL PROPERTIES OF COPPER SUBJECTED TO HIGH PRESSURE TORSION

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ABSTRACT

Experiments were conducted on copper subjected to High Pressure Torsion to investigate the evolution of microstructure and microhardness with shear strain, γ. Observations have been carried out in the longitudinal section for a proper demonstration of the structure morphology. An elongated dislocation cell/subgrain structure was observed at relatively low strain level. With increasing strain, the elongated subgrains transformed into elongated grains and finally into equiaxed grains with high angle grain boundaries. Measurements showed the hardness increases with increasing γ then tends to saturations when $\gamma > 5$. The variation tendency of microhardness with γ can be simulated by Voce-type equation.

Key words: copper, microstructure, microhardness, High Pressure Torsion

INTRODUCTION

High Pressure Torsion (HPT) is one of the effective severe plastic deformation methods to refine the grain size of alloys and pure metals with pore free nanostructure [1-3]. Transmission electron microscopy (TEM) has been used [1-3] to characterize the microstructural evolution during HPT straining, and observations of structural refinement were reported. But, it is noted that most of these studies have been devoted to an examination of the microstructural evolution in the torsion plane, and the observed deformation structure is complex but it was generally characterized as having equiaxed morphology [1-3]. However, the ideal section for microstructure observation in HPT processed samples is the longitudinal section (TD section) [4]. In the present study, microstructure observation in copper deformed by HPT was carried out in the TD section by TEM, and microhardness measurements were also conducted so as to analyze the variation tendency of microhardness with γ.

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EXPERIMENTAL PROCEDURE AND RESULTS

Copper disk with the diameter of 12 mm and the thicknesses of 3.5 mm was placed between upper and lower anvils of the HPT facility, as illustrated in Fig.1(a).
Then it was subjected to HPT under a pressure of 1.2 GPa at room temperature.

The microstructure of copper in TD section (Fig.1.(b)) after different HPT revolutions was examined by using JEOL100XL transmission electron microscope (TEM) operating at 100 kV. Selected area electron diffraction (SAED) was also performed under each condition from regions having diameters of 2.5 μm. Measurements of subgrain size and grain size were made directly from TEM photographs using the linear intercept method.

Vickers microhardness tests were also carried out on TD section, which contains the torsion axis of sample (Fig. 1(b)), with an MH-5L hardness tester using a load of 50g for 15 s. The shear strain, γ at a given r can be calculated according to $\gamma=2\pi Nr/h$, where N is the number of revolutions, h is thickness of sample and r is the distance from the center of sample (axis of rotation).

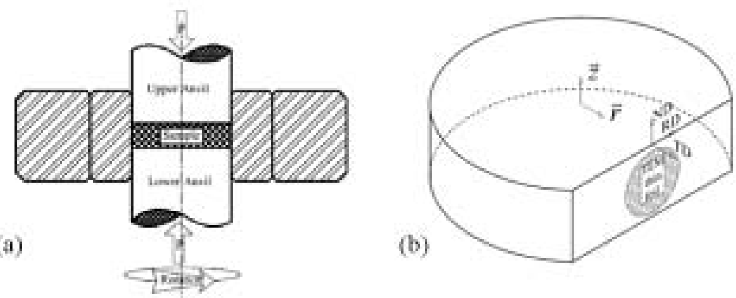

Figure 1. Schematic diagram of the device used for HPT (a) and sampling positions for microstructure observation and Vickers microhardness measurement (b). $\bar{z}$: torsion axis, $\bar{r}$: radius.

Fig. 2 shows the typical bright-field micrographs and their corresponding SAED patterns of Cu deformed by HPT with different shear strains. At γ =2.4 (Fig. 2(a)), elongated subgrains surrounded by parallel extended dislocation boundaries with an average spacing of 250 nm were observed. With increasing strain, the average spacing of the elongated subgrains decreased to 170 nm, and some elongated grains with sharp boundaries appeared (Fig.2(b)). Increasing shear strain to 21.5 (Fig. 2(c)), some grain boundaries with large orientation difference appeared, and most of the elongated subgrains transformed into equiaxed grains with an average grain size of 273 nm. The diffraction pattern also provided evidence for the existence of large-angle misorientations in the microstructure. At γ =37.3 (Fig. 2(d)) equiaxed microstructure with an average grain size of 285 nm were the dominated features of the deformed microstructure. It is evident from the SAED patterns that many of gains are separated by large-angle grain boundaries.

Fig.3(a) shows the measured microhardness of all samples as a function of the radius. The lower dotted line indicates a microhardness of Hv ≈52 recorded for the material in the initial annealed state. The lower set of experimental points, lying in the vicinity of Hv≈89, represents the measured hardness across the sample immediately after application of the load but without torsion-straining. All graphs show the same trend: smallest values at small radii (near the center of sample) and a continuously increase with increasing distance from the center and then tends to saturation at the outer region of the samples. The data of Fig. 3 (a) are re-plotted in Fig. 3(b) by using $\gamma = 2\pi Nr/h$. The results show that, within the relatively small extent of shear strain (0~5), the microhardness of copper specimens increases with the increase of shear strain and then tends to saturation at relatively large strain level.

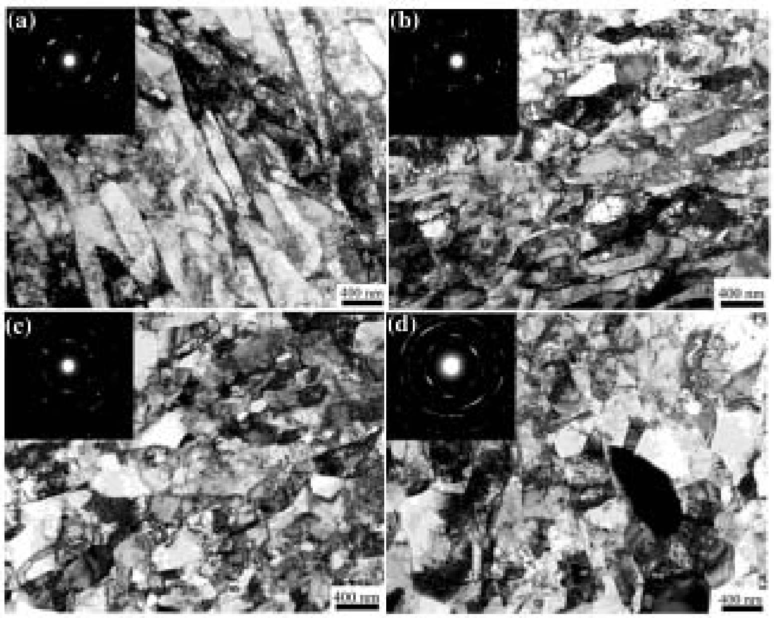

Figure 2. Bright-field TEM micrographs and SAED patterns in TD sections of Cu deformed by HPT with different shear strains: (a) γ =2.4 (N=1, r =1.35mm), (b) γ =12.9 (N=4, r =1.8 mm), (c) γ =21.5 (N=4, r =3 mm), and (d) γ =37.3 (N=16, r =1.3 mm).

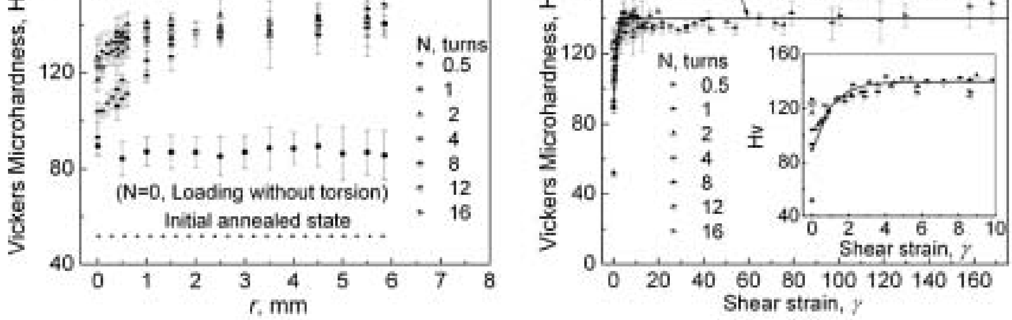

Figure 3. Measured Vickers microhardness as a function of radius (a) and shear strain (b) in copper samples subjected to HPT with different turns. The inset is an enlargement showing the distribution of the microhardness at relatively low strain level.

The variation tendency of microhardness with γ, is very similar to the variation tendency of stress with strain described by Voce type equation [5]

$$\sigma(\varepsilon) = \sigma_s - (\sigma_s - \sigma_0)\exp(-\varepsilon / \varepsilon_c) \qquad (1)$$

where σ_s is the saturation flow stress at high strains, and σ_0 and ε_c are constants dependent upon the material and the temperature. By adopting the approximate relationship between hardness, Hv, and flow stress, σ, as $H\text{v}=3\sigma$, the relationship between Hv and the imposed shear strain γ may be given as

$$H\text{v}(\gamma) = H\text{v}_s - (H\text{v}_s - H\text{v}_0)\exp(-\gamma/\gamma_c) \qquad (2)$$

where $H\text{v}_s$ is the saturation hardness at large strains, $H\text{v}_0$ represents the measured hardness of the sample immediately after application of the load but without torsion-staining, and γ_c is a constant dependent upon the material and the temperature. Constant $\gamma_c=1$ was used in the present calculation.

As indicated by the solid line in Fig. 3(b), the variation tendency of the calculated hardness according to Eq.(2) with γ agrees well with the measured results.

CONCLUSIONS

Microstructure observations in TD section revealed a rather different morphology of the dislocation structures from those reported in the previous studies [1-3]. An elongated dislocation cell/subgrain structure forms at low shear strain level. With increasing shear strain, the elongated subgrains transform into elongated grains and finally into equiaxed grains with high angle grain boundaries.

The microhardness increases dramatically near the center of the sample or within a relatively small extent of shear strain, then tends to saturation at the outer region of sample or at a relatively large strain. This tendency can be simulated by Voce-type equation.

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