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# Influence of Shear Banding on the Formation of Brass-type Textures in Polycrystalline fcc Metals with Low Stacking Fault Energy

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Texture evolution in nickel, copper and α-brass that are representative of face-centered-cubic (fcc) materials with different stacking fault energy (SFE) during cold rolling was systematically investigated. X-ray diffraction, scanning electron microscopy and electron backscatter diffraction techniques were employed to characterize microstructures and local orientation distributions of specimens at different thickness reductions. Besides, Taylor and Schmid factors of the {111} <110> slip systems and {111} <112> twin systems for some typical orientations were utilized to explore the relationship between texture evolution and deformation microstructures. It was found that in fcc metals with low SFE at large deformations, the copper-oriented grains rotated around the <110> crystallographic axis through the brass-R orientation to the Goss orientation, and finally toward the brass orientation. The initiation of shear banding was the dominant mechanism for the above-mentioned texture transition.

**KEY WORDS: Face-centered-cubic alloy; Texture; Rolling; Shear band**

## 1. Introduction

Texture is a common phenomenon in crystalline materials. It is closely linked to mechanical behaviors, physical and chemical properties of materials, and is associated with the performance in their production processing and follow-up engineering applications. Hence, an accurate understanding of deformation mechanisms together with precise control of texture evolution becomes a prerequisite for developing new materials[1]. As for face-centered-cubic (fcc) materials with high or medium stacking fault energy (SFE), grains tend to rotate toward the copper-type textures (inclusive of the {112} <111> and {123} <634> orientations) with increasing deformation, and dislocation slip has been identified as the dominant deformation mechanism during the whole deformation stage[2–4]. For materials with low SFE, the copper-oriented grains also aggregate at an early stage of deformation. However, with increasing loading these grains then rotate to the brass-type textures (inclusive of the {110} <112> and {110} <100> orientations) and the latter dominates in the final textures at large deformations[5–7].

Numerous efforts have been made to understand the texture evolution in low SFE fcc metals at large deformations[8,9]. According to the research on cold-rolled α-brass by Wassermann et al.[10], twinning of {112} <111> (copper) oriented materials introduces the twin orientation {552} <115> (copper$^{T}$), and then, the copper$^{T}$-oriented grains rotate toward the {110} <112> orientation (brass) by dislocation slip. However, Duggan et al.[11] clarified that the total volume fraction of twin is less than 25%. Therefore, the limited twin volume is not sufficient to change the texture type completely. Based on the study of microstructure for α-brass at large thickness reductions, shear bands are proposed to induce the significant texture change from the copper-type to the brass-type after the formation of the fine matrix-twin lamellar structure[11]. Besides, Hirsch et al.[8,12] developed a physical model by proposing that, at small deformations the copper$^{T}$-oriented grains which are developed by mechanical twinning rotate to the intermediate orientation of {111} <112> (brass-R), then, the brass-R grains rotate to the Goss orientation through the process of shear banding deformation. In addition, a series of researches on low SFE fcc single crystals initially oriented with copper have been conducted by Paul et al.[13,14]. It is suggested that two coplanar slip systems that are adjusted by the initiation of shear banding play an important role in the formation of the brass-type textures.

For the simulation of deformation textures in low SFE fcc metals, so far only a few studies have addressed incorporating shear band theory. A numerical study by Gill Sevillano et al.[15]

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concluded that shear bands in rolled fcc materials with low SFE do not play a significant role in the texture evolution. On the contrary, Kalidindi[16] conducted numerical simulations by introducing a specified shear component stemming from shear bands into the velocity gradient tensor for grains whose twin volume fraction reaches a critical value. He suggested that the mechanism of shear banding contributed to the copper-to-brass type texture transition. Therefore, although extensive studies in both experiments and simulations have been made[17,18], the mechanism of texture evolution in low SFE fcc metals at large deformations are still not clear.

In the present study, new features about the $\alpha$ fiber ($\Phi = 45°$, $\varphi_2 = 90°$) and the $\tau$ fiber ($\varphi_1 = 90°$, $\varphi_2 = 45°$) in the Euler space for fcc materials with different SFEs are studied. In order to explore the relationship between texture evolution and deformation mechanisms composing of dislocation slip, twinning and shear banding in metals with low SFE, the microstructures and local orientation distributions of the metals at different deformation stages were characterized by using EBSD techniques. Moreover, Taylor and Schmid factors for some typical orientations under plain strain state (cold-rolling) were analyzed[19,20]. This study provides an unambiguous understanding on the micromechanics of texture evolution in fcc metals with low SFE as well as a guidance for developing the relevant materials, such as austenitic stainless steels[21−23] and twinning-induced plasticity (TWIP)[24−26] steels incorporating fcc phases.

## 2. Experimental

Pure nickel (99.99%), copper (99.99%) and $\alpha$-brass (Cu−32% Zn) were selected as the studied materials in this work. They represent fcc materials with high, medium and low stacking fault energy (SFE), respectively. All the metals were firstly cross-rolled to a total reduction of 30% and heat treated to provide equiaxed microstructures with an average grain of about 30 μm[11], which were taken as the initial states of the materials. Then, the materials were symmetrically rolled at room temperature to different thickness reductions: 20%, 40%, 80% and 98%. The rolling processes were divided into many steps, so as to ensure homogeneous deformation and to avoid generating a large quantity of heat in each step. The {111}, {200} and {220} pole figures for the three materials with various reductions were measured by the laboratory X-ray diffraction (XRD) technique with a radiation of Cu$K\alpha$ in reflection geometry. Then, crystallographic orientation distribution functions (ODFs) for the materials were determined from their respective pole figures by using the LaboTex texture analysis software[28]. The notation of {*hkl*} <*uvw*> was used for characterizing the texture type with {*hkl*}||ND and <*uvw*>||RD, where RD represented the rolling direction and ND denoted the normal direction of the rolled plates. The microstructures and orientations of $\alpha$-brass at various deformation stages were characterized by employing the electron backscatter diffraction (EBSD) technique on the longitudinal section of each sample containing RD and ND. In order to obtain EBSD patterns of good quality, it is important to remove the residual deformation in the surface layer of specimens. The investigated surfaces (the longitudinal sections) were first mechanically ground and polished. Subsequently, electro-polishing was carried out for the flat surfaces in a solution of 20% phosphoric acid ($H_3PO_4$), 20% ethanol ($C_2H_5OH$), 10% 1-propanol ($CH_3CH_2CH_2OH$) and 1% urea ($((NH_2)_2OH)$) in volume at 20 °C, with the voltage of 15 V for 25 s. The EBSD mapping was performed in a field emission gun scanning electron microscope (JEOL JSM 7001F) with EBSD acquisition camera and channel 5 software. Beam control mode was used to acquire the Kikuchi pseudo-bands from crystals. An acquisition step of 0.02 μm was applied for the 80% and 98% deformed specimens and for other specimens a step size of 0.2 μm was used.

Taylor factors and Schmid factors for some typical orientations under plain strain state were calculated. In this study, the Taylor factors for the studied orientations were obtained by the full constraints Taylor model[19,27]. Besides, the Schmid factors for the twelve {111} <110> slip systems under plain strain state were computed for the typical orientations. Then, the maximum among the absolute values of the different systems was selected as the Schmid factor of each orientation. Using the same method,

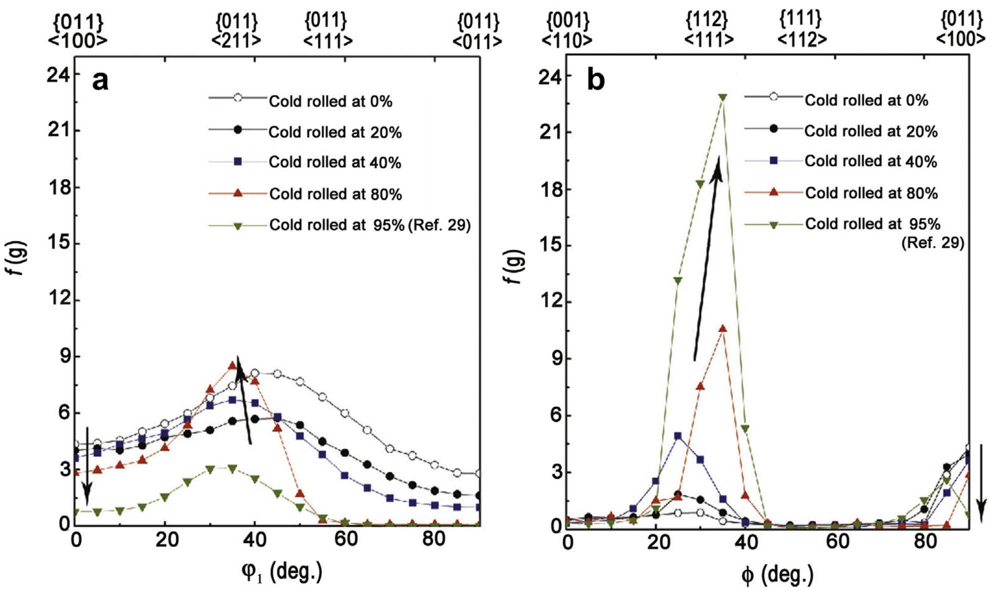


**Fig. 1** Texture evolution of nickel with rolling reduction in $\alpha$ and $\tau$ fiber representation: (a) $\alpha$ fiber with constant Euler angles $\Phi = 45°$ and $\varphi_2 = 0°$, (b) $\tau$ fiber with constant Euler angles $\varphi_1 = 90°$ and $\varphi_2 = 45°$. The data set at 95% deformation is obtained from Ref.[29] for nickel with the same initial microstructure as in this work.