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Effect of preferential orientation on the annealing twins during the low temperature treatment in nickel

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ABSTRACT

The influence of preferential orientation (namely cube texture) on annealing twins in a cold rolled (95% thickness reduction) pure nickel has been investigated during isothermal annealing at low temperature. The results reveal that the cube oriented grains in cube related twinning (CT) regions have larger average boundary migration rate in comparison with grains in non-cube related twinning (NCT) regions. However, the twin density is always lower in CT regions than that in NCT regions. This is contrary to the growth accident model that suggests that twin density was proportional to the grain boundary velocity and grain boundary distance. It can be interpreted by the fact of the selective nucleation and growth mechanism for cube oriented grains which are less dependent on annealing twinning during recrystallization. Furthermore, during grain growth the average length of annealing twin boundary segments has a significant increment, which is probably attributed to lengthening of low energy twin boundaries along with the migration of other neighboring boundaries.

1. Introduction

Annealing twins were initially reported by Carpenter and Tamura [1] in 1926, which were frequently observed in deformed and subsequently annealed face-centered-cubic (F.C.C.) materials with low and medium stacking fault energy (SFE). Subsequently, Gertsman et al. [2] revealed that annealing twins could significantly affect the microstructure and texture of the deformed and annealed alloys. Further, it was found that the annealing twins have a significant effect on materials properties, such as corrosion, creep resistance, and weldability [3]. Therefore, the researches on the evolution of annealing twins have become attractive for enhancing materials performance.

To account for the annealing twin formation, several mechanisms have been formulated, such as (a) the grain encounter model [4] (b) stacking fault model involving nucleation of twins by stacking faults or fault packets [5,6] and (c) growth accident model [7–10] suggesting that a twin boundary formed at a migrating grain boundary due to a stacking error. However, there was no well agreement on the formation of annealing twins, although the growth accident model was widely accepted by a majority of researchers. Based on growth accident model, some researchers had attempted to predict the twin density which was related with the microstructural and thermo-mechanical parameters including grain size, stacking fault energy, deformation reduction and temperature. Gleiter [7] suggested that the twin density was determined by both grain size and temperature. Nevertheless, Pande [8] derived a formula that indicated that the twin density was uniquely dependent on grain size, and the temperature played a role just through grain size dependence. And most recently, some studies were carried out to analyze the twin formation during recrystallization process, with the purpose of enriching the mechanism of annealing twin formation. Wang et al. [11] draw a conclusion that the level of stored energy played a decided role on the twin density during the stage of primary recrystallization. Furthermore, Jin et al. [12] clarified the influence of local stored energy variation on twin density and concluded that the higher the recrystallization front tortuosity was, the higher the twin density was. Unfortunately, little available information about the role of crystallographic texture in annealing twin development was given.

For metal substrate materials used in YBCO-coated conductor, crystallographic texture is a significant consideration to improve weak-link effect, and unfortunately strong cube texture was generally weakened by the presence of annealing twins which have an exact relationship 60° $< 111 >$ with cube oriented grains. Thus, annealing twins have an adverse effect on strong cube texture, owing to that annealing twins always lead to randomization of texture [2]. Therefore, it is necessary to elucidate how the annealing twins evolve accompanied by the formation of cube texture and furthermore understand the chance that annealing twin occurs in grains with different orientations. Recently, Field et al.

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[13] analyzed that annealing twins played a key role when grain growth stagnated during recrystallization in highly deformed copper but did not involve texture. Besides that, one of our authors Chen [14] suggested that the formation of cube texture could promote the formation of annealing twins in high deformed nickel as a function of annealing temperature during recrystallization. Whether the occurrence of annealing twins is closely correlated with preferential orientation during recrystallization? For this reason, high deformed ($\varepsilon_{vm}=3.46$) pure nickel is performed isothermal low temperature annealing, which not only obtains relatively strong cube texture but on the other hand also slows down the recrystallization process to capture more details.

The main objective of the present work is to identify the influence of preferential orientation on annealing twin density during the recrystallization and grain growth in heavily deformed pure nickel, which may lead to a better understanding of relationship between cube orientation and annealing twin boundaries network in strongly textured materials.

2. Experimental materials and procedures

The as-received hot-rolled pure nickel (99.95%) was cold-rolled to 50% in thickness reduction, subsequently followed by annealing at 600 °C for 30 min with Ar-4%H_2 protective atmosphere in order to randomize texture in the initial condition. Then unidirectional and multi-pass cold rolling with 10% reduction per pass was carried out on the homogenized material with a total thickness reduction of 95%. The highly deformed specimens were annealed at 550 °C for different durations to capture the early stage of recrystallization in a tube furnace with protective atmosphere of Ar-4%H_2 and then quenched in cold water immediately.

The microstructures and micro-texture of the deformed and annealed specimens were characterized by electron backscatter diffraction (EBSD) system attached to TESCAN MIRA 3 field emission gun-scanning electron microscope. The longitudinal section of the specimens (plane containing normal direction (ND) and rolling direction (RD)) was mechanically and electrolytically polished for the EBSD characterization. A step size of one-tenth of the average grain size was used and at least three maps were taken for each condition with RD as the horizontal direction in all EBSD orientations maps. Due to the limited angular resolution of the EBSD technique [15], a critical misorientation angle of 2° was defined to identify boundaries in the orientation maps, where low angle grain boundaries (LAGBs) and high angle grain boundaries (HAGBs) were identified by misorientations of 2–15° and > 15°, respectively. Annealing twin boundaries were defined by a misorientation of 60° with a tolerance of 8.66°, according to Brandon's criterion $\Delta\theta_{max}=15°\Sigma^{-1/2}$ [16], regardless of their coherent versus incoherent character. The average grain size was calculated without considering twin boundaries.

In order to quantify annealing twins, the annealing twin density (N_L) was used for evaluation, which was defined as

$$N_L = \frac{L_{tb}}{S} \times \frac{2}{\pi}$$

where L_{tb} is the length of twin boundary detected in a given sample section area S [12]. Note that in the present work the annealing twin density was determined only in the regions of recrystallized grains. The recrystallized grains were extracted from the EBSD maps based on the criterion shown in [17].

Macro-textures of the samples before and after annealing were measured by X-ray diffraction (XRD, Rigaku D/max-2500 PC). The volume fractions of different texture components were determined within a spread of 15° around their respective ideal locations in Euler space by Labotex, where {001} < 100 > (Cube), {122} < 212 > (Cube-T), {110} < 112 > (Brass), {123} < 634 > (S), {112} < 111 > (Copper), and {011} < 100 > (Goss) were considered.

To evaluate the recrystallization process, the Vickers hardness measurements were performed on each sample using a load of 300 g with a dwell time of 15 s. In order to ensure the statistical analysis, at least 10 indentations were made for each specimen to obtain an average value.

3. Results

3.1. Micro-hardness

Fig. 1 depicts the micro-hardness evolution of samples annealed at 550 °C for different holding times. After 5 min annealing there is an obvious drop in the hardness curve, suggesting the occurrence of recovery or the onset of recrystallization. Subsequently, the hardness value decreases slowly with increasing annealing time. When samples were annealed for 30 min, a dramatic decrease in the curve is observed. After that, the variation in hardness is in a small fluctuation, evidencing that the deformed microstructure has been substituted by dislocation-free recrystallized microstructure.

3.2. Macro-texture evolution

The purpose of the global texture measurement is to gain an overall idea about the deformation and recrystallization textures from reasonably large sample areas. The texture evolution of deformed and subsequently annealed samples is displayed in the sections of $\varphi_2=0°$, 45° and 65° of orientation distribution functions (ODFs). The ODFs of cold-rolling sample in Fig. 2(a) are indicative of the typical "copper-type" rolling texture, namely, Brass, Copper and S. After annealing for 20 min, the microtexture is mainly composed of rolling texture and recrystallized cube component, which suggests that the sample has been partially recrystallized. Further annealing to 30 min, the cube orientation gets enhanced, while residual deformation texture is extremely weak, as depicted in Fig. 2c. Extending annealing time to 300 min, the texture is distinctly dominated by cube orientation along with weak cube-twin orientation, as evidenced in Fig. 2d. The quantitative analysis of texture component is presented in Fig. 3. It can be observed that there is no obvious changes in macro-texture at the initial stage of annealing, indicating that the slow process of recrystallization at low temperature. After annealing for 20 min, the volume fraction of cube texture increases rapidly, meanwhile

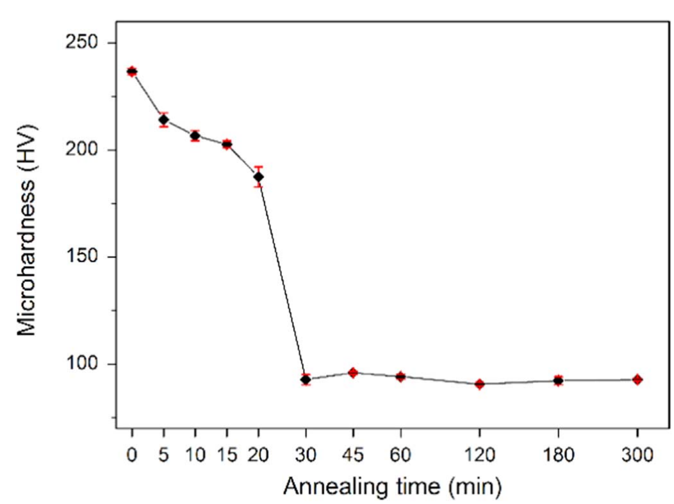

Fig. 1. Evolution of micro-hardness as a function of annealing time.