New Insight into the development of deformation texture in face-centered cubic material

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Abstract. Despite a tremendous amount of research to understand the evolution of microstructural and texture changes during deformation, there is still a need for further deep insight into it. The most used deformation process under use in industries is the rolling of material. Stacking fault energy is the most critical material property, which governs the deformation behavior. In the present study, the microstructure and texture changes of a high SFE material (pure nickel) during cold rolling was investigated. The objective is to find criteria for the change of orientation of a particular grain as rolling proceed. Pseudo-in-situ rolling was done through which the shape, orientations, and neighbors of each grain could be tracked as a function of rolling reductions. It was found that orientations of some grains remain stable after a large reduction and do not develop sub-grains within it, whereas orientations of some grains change drastically at a very low deformation level along with the evolution of a large number of sub-grains.

1. Introduction

The underlying deformation mechanisms and the corresponding development of microstructure and texture in face-centered cubic (FCC) material have been the subject of interest for several decades [1–4]. Plastic deformation in materials takes place mostly by either of two mechanisms, slip or twinning. After large deformation, when dislocation density in the material becomes very high, the mobility of these dislocations becomes difficult, which produces deformation heterogeneities like micro-bands, shear bands, etc. [1-3]. Along with it, as the deformation proceeds, the material attains a certain unique crystallographic texture through grain rotation and to accommodate the strain incompatibility developed during deformation [4]. Dislocation mobility and texture formation are inter-related and affected by the stacking fault energy (SFE) of material, the initial orientation of grains, type of deformation (Schmid factor of the grain), and misorientation across its neighborhoods to maintain strain incompatibility [4-7].

Rolling texture in high SFE FCC material such as Al, Ni etc. shows Cu-type texture having majorly Cu {112}<111>, Bs {110}<112> and, S {123}<634> components [4,8]. The texture of low SFE materials, known as Brass type texture, is characterized by the absence of Cu {112}<111> and presence of strong Bs {110}<112> and Goss {110}<001> components [4,8]. However, even in the case where deformation texture is considered to be very strong, more than 50% of grains belong to some random orientation. Most of the existing theory deals only with the formation of overall bulk texture and does not clearly explain the probability and the amount of rotation of a particular grain as a function of deformation. There exist certain criteria for such as the Schmid factor and Taylor factor, which deals with the probability of occurrence of plastic deformation in grain [9, 10]. However, it fails to explain the role of the neighboring grain orientation.

The earliest model that exists to predict the deformation texture is the Sachs model [9], which assumes that each grain of a polycrystalline material will behave as an unconstrained single crystal of the same orientation, i.e., independent of its neighboring grains. It also assumes a single slip operating in each grain. Because of all these assumptions, this model fails to predict the amount of rotation of each grain during deformation. The other model that exists is the full constraint Taylor model [10], which assumes that, during deformation, all the grains will undergo the same shape change as the entire
polycrystal. The other assumption of this model is that critical resolved shear stress for slip is the same for all the active slip systems, which strain harden at the same rate. The main shortcomings of Taylor’s model are that it neglects elastic strains and strain heterogeneities and that its prediction of the existence of at least five slip systems, which may not be true. Due to these assumptions, this model fails to predict texture changes in low SFE material. Recently developed, visco-plastic self-consisted (VPSC) model predicts deformation texture in large no. of cases [11, 12]. However, it does not incorporate intra-granular misorientation, different microstructural features, and predicts overall bulk-texture.

The aim of the present work is to understand the probability and the amount of rotation at the individual grain level as a function of deformation. Pseudo-in-situ rolling was performed through which the shape, orientations, and neighbors of each individual grain could be tracked as a function of rolling reductions. The role of initial grain orientation, Schmid factor, and neighbor grain orientation during deformation have been investigated.

2. Experimental details

Pure nickel of 99.99% purity granules was melted in a vacuum arc melting furnace. To break the as-cast dendritic structure and to randomize the initial solidification texture, materials were 50% cross-rolled and further annealed in vacuum at 800 °C for 6 hours. It produced a homogenized microstructure having strain-free equiaxed grains and a weak texture. Further, cold rolling of the material was done till 30% cold-rolling reduction. EBSD scan of initial and 10%, 20%, and 30% cold-rolled (CR) material were done at the same region in the material without intermediate polishing so that the same grain as a function of deformation could be tracked, called pseudo-in-situ rolling. EBSD scan was taken on the transverse plane (RD-ND) of the specimen in a field emission scanning electron microscope (SEM), equipped with electron backscattered diffraction (EBSD) system. Samples for EBSD scans were metallographically polished and then electro-polished using A2 solution. For EBSD scans, the SEM was operated at 25 kV, and a step size of 200 nm was used throughout for all the samples. TSL OIM 8.1 software was used for data analysis.

3. Results and discussions

Figure 1a-1d shows the IPF map of initial pure nickel and differently rolled materials scanned in the same region. The change in shape and orientation of some grains are shown by the arrow.
Figure 1: Inverse pole figure (IPF) map of pure nickel (a) Initial, (b) 10% CR, (c) 20% CR, (d) 30% CR, material, and corresponding (111) pole figure of (e) Initial, (f) 10% CR, (g) 20% CR, (h) 30% CR, material. Same regions were tracked during rolling without intermediate polishing so that change in morphology and orientation of individual grain could be captured. Tracking of some grains circled in red color is shown by the arrow. The IPF color coding refers to a direction parallel to ND.

It can be seen that some of the grains (region 1) have shown a negligible change in orientation (2°/0 1 -5°) even after 30% cold-rolling, whereas some grains (region 2) have shown a significant change in orientation (7°/1 1 3°) just after 10% CR. Similarly, the change in orientation of each grain can be calculated as a function of cold-rolling reduction. Figure 1e-1h shows (111) pole figure of initial and cold rolled material. It can be seen that as rolling proceeds, the Cu-type rolling texture is developing. However, the maximum intensity is quite low after 30% CR showing a value of 2.6.
Figure 2: Grain boundary character distribution (GBCD) map of pure nickel (a) Initial, (b) 10% CR, (c) 20% CR, (d) 30% CR, material, and schmid factor map of (e) Ni initial, (f) Ni 30% CR, material. The black circle shows a region of sparse LAGB, and the blue circle shows a region of dense LAGB.

For a better understanding of the change in orientation in different grains, grain boundary character distribution was plotted. Figure 2a-2d shows the GBCD map of initial and cold-rolled nickel. Initial material has shown mostly high angle grain boundaries (HAGB) (red) and having a grain size of about 15-20 μm. As the rolling proceeds, the fraction of low angle grain boundaries (LAGB) (blue) increases, and it shows the highest fraction in 30% CR material. However, it can be seen that the distribution of LAGB is not homogeneous. Some grains (region 1) do not show any sub-grain, and some grains (region 2) show dense sub-grains within the grain. It can be attributed to the ease of deformation of a particular grain, i.e., initial orientation and the Taylor/Schmid factor of the grain. Figure 2e and 2f show the Schmid factor map of initial and 30% CR material. Region 1 shows a higher Schmid factor (black circle) compared to region 2 (blue circle), indicating easier deformation in the former case and thereby less dislocation accumulation, which results in a small change in orientation as a function of deformation.

To further analyze the data, grain 1 and grain 2 was partitioned and shown in figure 3. Figure 3b and 3c clearly shows that grain 1 has developed negligible sub-grain and have orientation \{001\}<uvw>, whereas grain 2 has developed very dense sub-grain and have orientation \{110\}<uvw>. It was also observed that other grains of this family have also shown similar trends. However, a deeper
Insight of grain 1 shows that it has also shown some amount of sub-grains mostly near the edge of the grains, which indicates that formation of sub-grain within a grain does not only depend on its own orientation or Schmid factor but also depends on its neighboring grains orientation or misorientation across the grain boundaries. So, it can be concluded that the formation of sub-grain within the grain depends on its own orientation, i.e., Schmid factor of grain. However, the formation of sub-grain at the edge of grain may result from strain incompatibility at the grain boundary due to high misorientation across the interface [4-5].

Figure 3: Grain boundary character distribution (GBCD) map of (a) Ni 30%CR (b) grain having sparse LAGB, (c) grain having dense LAGB, (d) (001) pole figure of grain 1 showing {001}<uvw> orientation, and (e) (001) pole figure of grain 2 showing {110}<uvw> orientation.

To further investigate the role of neighboring grain orientation, misorientation across all the neighbors of grain 1 and grain 2 is plotted for initial material before rolling, in figure 4a and 4b, respectively. Grain 1 has five neighboring grains having an average misorientation of 54.6° across all of its neighbors. Grain 2 has eight neighboring grains having an average misorientation of 51.2° across all of its neighbors, which is almost the same as grain 1. It was seen earlier that grain 2 had shown a significant change in orientation along with the development of sub-grains within it, compared to grain 1. It can be concluded that numbers of neighboring grains are a more important factor than average misorientation across all the grain boundaries. Figures 4c and 4d show the Schmid factor map of grain 1 and grain 2, along with all of its neighbors, before rolling. It can be seen that grain 1 has a higher Schmid factor compared to grain 2. However, the Schmid factor difference in grain 1 across all of its neighbors is less compared to grain 2, which can also play an important role during the rotation of grain during deformation. High Schmid factor difference across the grain boundary may result in high strain incompatibility at the interface and it can promote more rotation of the grain, which can be seen in case of grain 2.
4. Conclusions

Pseudo-in-situ rolling of pure nickel was performed to investigate the role of different factors in the change of orientation of individual grain as a function of rolling. The following conclusion could be drawn from this study.

➢ {110}<uvw> orientation has shown a greater change in orientation and development of sub-grain within it compared to {100}<uvw> orientation.
➢ The number of neighboring grains and Schmid factor of grain along with the Schmid factor of neighboring grain play an important role in the rotation of grain during deformation.

Figure 4: IPF map of two selected grain in nickel before rolling (a) grain 1 along with its neighbors, (b) grain 2 along with its neighbors, (c), (d) corresponding Schmid factor map of the same region, respectively. (e), (f) IPF map of the same region through tracking all the grains after 30% cold rolling, respectively. The white line shows a misorientation angle across the grain with all of its neighbors.
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