Effect of strain path change on the texture evolution after cold rolling and recrystallization of Nickel-40 wt. % Cobalt alloy

Gyan Shankar, Satyam Suwas

Department of Materials Engineering, Indian Institute of Science, Bangalore-560 012, India

Email: gyanshankar@iisc.ac.in

Abstract

Effect of change in strain path during rolling and its role in the development of crystallographic texture after deformation and then after recrystallization has been studied for Ni-40wt%Co which is a medium stacking fault energy (SFE) material. Results indicate that textures developed after unidirectional rolling and multi-step cross-rolling are quite different which also leads to different texture evolution after recrystallization. Microstructure of both UDR and cross rolled sample shows band type of feature within the grain having lesser band density for cross rolled samples. After recrystallization unidirectionally rolled sample shows bimodal whereas cross rolled sample shows equiaxed grain size distribution. But all the samples show large fraction of annealing twins. It has been shown that in all cases formation of annealing twins have large influence on the texture transition after recrystallization.

Keywords: Deformation texture; Cold rolling, Unidirectional rolling; Cross-rolling; Texture transition; Recrystallization texture; annealing twin

1. Introduction

The mechanisms of evolution of deformation texture in face centred cubic (FCC) metals and alloys have been investigated by many researchers [1–8]. One of the most important material property that controls the micro-mechanisms of deformation in FCC materials is the stacking fault energy (SFE). It is well known that rolling texture in a closed packed face centred cubic materials show texture transition from Cu (or pure metal) type to Bs (or alloy type) with a decrease in stacking fault energy. In general, high and medium SFE material shows Cu type texture which consist of prominent Cu {112} <111> component, a continuous spread of orientation from Bs {110} <112> to S {123} <634> and a weak Goss {110} <001> components. Low SFE material shows Bs type texture which consists of strong Bs {110} <112> and Goss {110} <001> components. This transition in deformation texture with a change in SFE is also accompanied by deformation heterogeneities such as deformation twins, microbands and shear bands. In a low SFE material, it was found that twinning occurs in the initial stage of deformation process, at the intermediate level twinning activity ceases and later at high strain formation of shear bands take place. Experiments conducted on single crystals revealed that Cu orientation is very prone to shear banding, whereas Goss orientation is not susceptible to heterogeneities during deformation [9, 10].

Apart from the material property, the processing variables are also equally important in the formation and stability of a particular texture [5, 11, 12]. It was found that change in strain path during rolling has pronounced effect on the evolution of texture [13–15]. Rotation of the
sample after every pass during rolling rotates the substructure along the new rolling direction. It leads to destabilizations of the substructures formed during the previous pass of rolling and thereby alters the evolution of texture.

Formation of recrystallization texture strongly depends on the initial deformation texture which is dependent upon the stacking fault energy of material and the processing parameters of deformation such as, mode of deformation, strain, strain path etc. Previous investigations on the development of recrystallization texture show that Cu type rolling texture typically results in cube recrystallization texture and Bs type rolling texture usually recrystallize in brass orientation [16]. Annealing twin is also an important factor which contributes to the formation of nuclei during early stages of recrystallization and later development of texture which was not present in the initial deformed material [17].

The aim of the present work is to understand the effect of strain path on the development of texture during deformation and subsequent recrystallization in Ni-40wt. %Co alloy. Ni-Co alloys form completely substitutional FCC solid solution up to 65 wt. %Cobalt. Stacking fault energy of the system decreases with cobalt addition from high SFE of (pure Ni, 150 mJ m$^{-2}$) to very low SFE of (Ni-60Co, 30 mJ m$^{-2}$) [18]. Ni-40Co alloy, used in the present work is a medium SFE material with SFE value is about 60 mJ m$^{-2}$. The micro-mechanisms of texture evolution in Ni-40Co are elaborated in this investigation.

2. Experimental Procedure

An ingot of nickel-cobalt alloy containing 40 wt. %Co was melted in vacuum arc melting furnace. During melting magnetic stirring was used to ensure composition uniformity. Quantitative composition analysis was carried out by energy dispersive spectroscopy (EDS). To break the cast dendritic structure and to weaken the initial solidification texture, the as-cast ingot was cross-rolled to 50% thickness reduction. After that, the rolled material was further annealed at 700 °C for 6 hours, to get completely homogenised strain free equiaxed microstructure having random texture. It was taken as starting material for our study. The initial material was then subjected to 70% unidirectional rolling (UDR) and multistep cross rolling (CR), equivalent to a true strain of ~1.2 at room temperature. A schematic representation of rolling schedules followed in the present study is shown in Fig.1. Both the Unidirectional and the cross rolled material was then subjected to isothermal annealing at 550 °C for 30 minutes for recrystallization.
Fig. 1. A schematic representation of rolling pass followed in the present study. Rolling direction in each pass is shown by arrow. 0.1 true strain has been given in each pass of rolling.

Bulk texture measurements for all the rolled and recrystallized samples were carried out by the Schulz reflection method, using an X-ray texture goniometer with Co Kα radiation (D8 Discover, Bruker AXS, Germany). Four incomplete pole figures (α = 0°-75°) were measured from the 111, 200, 220 and 311 peaks on the rolling plane of the sample. Because of coarse grain for initial sample texture measurements were done at multiple locations to improve the statistics of the measurements. The three-dimensional orientation distribution function (ODFs) was calculated using Labotex® software [19].

The microstructure of all rolled and recrystallized samples was taken in a field emission scanning electron microscope, equipped with EBSD system. All microstructure were taken on the transverse plane of rolled samples. EBSD data has been rotated to ND-plane to see the important texture component. Samples for EBSD scans were mechanically polished using silicon carbide emery sheets and further electropolished.

3. Results and Discussion

Fig. 2 shows (111) pole figures for all the starting, deformed and recrystallized Ni-40wt. %Co alloy. A key pole figure depicting the important texture components is also presented along with. The Miller indices of ideal FCC rolling and recrystallized texture component are given in Table 1.
Fig. 2. The (111) pole figures for (a) Initial material, (b) 70% UDR, (c) 70% UDR and recrystallized, (d) A key pole figure showing all important texture components, see table 1 for miller indices of these components (e) 70% CR, (f) 70% CR and recrystallized.

The starting material shows random texture because of prior cross rolling and annealing as mentioned earlier. After 70% UDR, it shows the development of strong Bs and Goss component with weak Cu and S component. It also shows some amount of Goss twin orientation which developed as a result of unidirectional rolling. On the other hand, the 70% Cross rolled sample shows weaker texture as compared to UDR which is also reported by many authors [20]. It shows remarkable Goss and {110} <111> (A) component. The development of ‘A’ component after cross rolling has also been reported for copper [20]. The intensity of Cu and S component was found very weak as compared to UDR sample. Cross rolled samples also show retention of Bs, G/B and rotated Goss component which was present in the initial material. The volume fraction of important deformation and recrystallization texture component was calculated after
computing the orientation distribution functions. The volume fractions of different texture components are shown in Fig. 3.

Table 1. Ideal FCC rolling and recrystallization texture components.

| Texture component         | Miller indices | Eular angle (Bunge) (φ1 φ2 φ3) |
|---------------------------|----------------|---------------------------------|
| Cu                        | {1 1 2}        | <1 1 1>                         |
| Bs                        | {1 1 0}        | <1 1 2>                         |
| S                         | {1 2 3}        | <6 3 4>                         |
| Goss                      | {1 1 0}        | <0 0 1>                         |
| Cube                      | {0 0 1}        | <1 0 0>                         |
| Taylor                    | {4 4 11}       | <13 11 8>                       |
| R                         | {1 2 3}        | <4 1 2>                         |
| G/B                       | {1 1 0}        | <1 1 5>                         |
| Rotated Goss (Rt-G)       | {1 1 0}        | <0 1 1>                         |
| Rotated Cu (Rt-Cu)        | {1 1 2}        | <1 1 0>                         |
| Rotated Cube (Rt-C)       | {0 0 1}        | <1 1 0>                         |
| Goss Twin (GT)            | {1 1 3}        | <3 3 2>                         |
| Cu Twin (CuT)             | {5 5 2}        | <1 1 5>                         |
| A                         | {1 1 0}        | <1 1 1>                         |
| F                         | {1 1 1}        | <1 1 2>                         |
| E                         | {1 1 1}        | <0 1 1>                         |
| α - fibre                 | {1 1 0}        | <0 0 1> - {1 1 0} <1 1 2>       |
| β - fibre                 | {1 1 2} <1 1 1> - {1 2 3} <6 3 4> - {1 1 0} <1 1 2> |
| τ - fibre                 | {1 1 2} <1 1 1> - {5 5 2} <1 1 5> - {1 1 0} <0 0 1> |
| Y - fibre                 | {1 1 1} || ND ((1 1 1) <0 1 1> - (1 1 1) <1 1 2>) |

After recrystallization of 70% UDR sample shows strong Brass, cube, Goss and Rt-cube component in almost equal proportion. It can also be seen that the intensity of Cu, Bs, S and Goss component went down as compared to deformed samples whereas the intensity of cube and Rt-cube component got increased. The 70% Cross rolled and recrystallized sample shows completely different texture as compared to UDR recrystallized samples. The development of strong rotated Goss component is clearly noticed, which was not present in the UDR recrystallized sample. It also shows little increases in Cu and S component as compared to cross rolled sample which was decreasing when UDR sample was recrystallized. One similarity were found in texture transition during recrystallization of both UDR and cross rolled sample is that Goss component always decreases after recrystallization.

To get the better understanding of texture transition upon rolling and after recrystallization, important fibres were plotted from the ODFs. The fibre plots are shown in Fig. 4. The α and τ- fibres show orientations <110> || ND and <110> || TD respectively. The α and τ- fibres are quite inhomogeneous. The β- fibre also does not show much variation in differently processed samples except for 70% cross rolled sample which exhibits a maximum between S/Bs and Bs orientation. The 70% UDR sample shows very high intensity near Bs and Goss orientation in α- fibre plot but after recrystallization, their intensity decreases. However, it maintains a broad peak between G/Bs and Bs orientation. The retention of some Brass orientation after recrystallization can be explained by nucleation with orientation close to those
of deformed matrix. It can take place by twinning into crystallographic identical variants influenced by annealing twin since lots of annealing twins can be observed in the microstructure after recrystallization (Fig. 6c) [21]. The 70% UDR recrystallized sample also shows a remarkable peak between CuT and Goss orientation with some intensity at Rt-Cube orientation in τ-fibre which cannot be ignored. The presence of CuT orientation could be attributed to the annealing twin of Cu orientation which was present in the deformed matrix. The development of rotated cube (Rt-C) component may take place by transformation of CuT orientation since it possesses $\Sigma 3$ twin relation with it.

![Fig. 3.](image)

Fig. 3. The volume fraction of important texture component of initial rolled and recrystallized material.

On the other hand, 70% cross rolled sample shows a strong peak near Goss and {110} $\langle111\rangle$ (A) orientation in $\alpha$- fibre and very high intensity between S/Bs and Brass orientation in $\beta$-fibre. The reason behind the occurrence of high intensity between S/Bs and Brass orientation could be destabilization of substructure because of rotation of sample after subsequent passes of rolling. Since Ni40Co is a medium SFE material, initial unidirectional passes leads to develop some Brass orientation but because of rotation of sample previous orientation destabilize and try to rotate towards other very preferential rolling texture component S orientation. As a result finally, high intensity develops at a location between S/Bs and Brass orientations which can be seen in $\beta$-fibre (Fig 4c). After recrystallization of 70% cross rolled sample it shows high intensity near brass orientation and a maximum at rotated Goss (Rt-G) orientation in $\alpha$-fibre. It can also be seen that there is a sharp decrease in Goss orientation after recrystallization. However, near Bs orientated grains recrystallized in the same manner as explained earlier for the unidirectional rolled samples through the formation of annealing twin,
since recrystallized cross rolled sample also shows a large fraction of annealing twins. Overall it can be observed that the formation of annealing twins has a large influence on the texture transition during recrystallization.

Fig. 4. Texture fibre plot of the initial Ni-40Co alloy, after 70% UDR, CR and after recrystallization (a) $\alpha$ – fibre, (b) $\tau$ – fibre and (c) $\beta$ – fibre.

The inverse pole figure (IPF) map for initial and all differently processed samples are shown in Fig.5. The initial microstructure shows a grain size of about 350 microns and consists of some annealing twins which was formed during homogenisation (Fig. 5a). After 70% unidirectional rolling microstructure shows lots of band type of feature inside the grain (Fig. 5b). These banded features are similar to the ones observed earlier in Al-Mg alloy by Hurley and Humphreys [22, 23], and later also in Ni40Co alloy reported by Madhvan [18]. They called these bands as “micro-shear bands”, because of its microscopic nature and confined within a single grain.
4. Conclusions

The effect of strain path on the evolution of deformation and recrystallization texture, as well as microstructure in a medium-SFE Ni–40Co alloy, has been investigated for 70% rolling reduction equivalent to a true strain of 1.2. The alloy showed quite different bulk texture and microstructure of UDR and cross rolled sample. Even after recrystallization, the effect of strain path during deformation can be seen in the formation of recrystallization texture. Marked differences from other medium-SFE materials, e.g. Cu, is clearly noticed. The important conclusions of this investigation are:

1. Unidirectionally rolled sample shows extended α- fibre having strong Bs and Goss component with some Goss twin orientation whereas cross rolled sample shows weaker texture having remarkable Goss and {110} <111> (A) component. Cross rolled sample also shows retention of Bs and rotated Goss component.

2. After recrystallization strong Bs, Goss, cube and Rt-cube orientation develop in almost equal proportion for UDR sample whereas strong rotated Goss component with little increase in Cu and S orientation formed for cross rolled samples.
3. For differently processed sample $\alpha$ and $\tau$-fibres are quite inhomogeneous but $\beta$-fibre shows does not show much variation except for 70% cross rolled sample showing a maximum between S/Bs and Bs orientation.

Acknowledgements

The authors acknowledge the extensive use of microscopy facilities at the Advanced Facility for Microscopy and Microanalysis (AFMM), Department of materials engineering, Indian Institute of Science.

References

1. Duggan BJ, Hatherly M, Hutchinson WB, Wakefield PT (1978) Met Sci 12:343–351.
2. Hutchinson WB, Duggan BJ, Hatherly M (1979) Met Technol 6:398–403.
3. Suwas S, Singh AK, Rao KN, et al (2011) Acta Metall 36:4827–4840.
4. LEE CS, DUGGAN BJ, SMALLMAN RE (1993) J Phys IV 3:2027–2032.
5. Hirsch J, Lücke K (1988) Acta Metall 36:2863–2882.
6. C. D, Valle R, Dervin P, Penelle R (1989) Acta Met 37:1547–1571.
7. Engler O, Kong XW, Lücke K (1999) Scr Mater 41:493–503.
8. Engler O (2000) Acta Metall 48:4827–4840.
9. Nakayama Y, Morii K (1987) Acta Metall 35:1747–1755.
10. Nakayama Y (1985) Acta Metall 33:379–386.
11. J. HIRSCHE KL and MH (1988) Acta Metall 36:2905–2927.
12. Engler O, Hirsch J, Cke KLI (1995) Acta Metall 43:121–138.
13. Suwas S, Singh AK (2003 Mater Sci Eng A 368–371.
14. Gurao NP, Sethuraman S, Suwas S (2011) Mater Sci Eng A 528:7739–7750.
15. Madhavan R, Ray RK, Suwas S (2014) Acta Mater 74:151–164.
16. K. Pawlik (1986) phys. stat. 801. (b) 134, 477.
17. Gurao NP, Kapoor R, Suwas S (2010) Metall Mater Trans A 41:2794–2804.
18. Saleh AA, Pereloma E V, Gazder AA (2011) Mater Sci Eng A 528:4537–4549.
19. Hurley PJ, Humphreys FJ (2003) Acta Mater 51:1087–1102.
23. Hurley PJ, Humphreys FJ (2003) Acta Mater 51:3779–3793.
24. Madhavan R, Ray RK, Suwas S (2014) Acta Mater 78:222–235.
25. Hansen N, Jensen DJ (1999) Philos Trans R Soc A Math Phys Eng Sci 357:1447–1469.
26. Garg R, Ranganathan S, Suwas S (2010 Mater Sci Eng A 527:4582–4592.