# A Model of Dynamic Recrystallization in Alloys during High Strain Plastic Deformation

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Recrystallized grains, less than 200 nm in diameter were observed in heavily shear zones of a high strength low alloy steel and a Ni-based alloy, and Also grain refinement, less than 3 $\mu$m in diameter was made in high purity aluminum by ECAE at ambient temperature. The experimental results showed that high strain rate and large deformation could induce dynamic recrystallization. Based on dislocation dynamics and grain orientation change enhanced by plastic deformation, a model for the recrystallization process is developed. The model is used to explain the ultra fine grains which are formed at a temperature still much lower than that for the conventional recrystallization.

## 1. Introduction

Very small recrystallized grains were observed in the adiabatic shear bands in metals when they were heavily deformed at high strain rates, which were zones of highly localized plastic deformation[1~5]. Pak[1] found that shear bands in a commercially pure titanium were consisted of small grains of 50~300 nm in diameter with well-defined boundaries. Chokshi[2], Andrade[3] and Hines[4] also observed the recrystallized grains with 100~200 nm diameters within the shear bands of copper. Cho[5] showed that there were equiaxed cells with the sizes of 200~500 nm in the shear band of HY-100 steel during dynamic torsional deformation. Recently, ultra-fine grained metallic materials were obtained using equal channel angular extrusion (ECAE), in which an intense plastic strain was imposed upon a polycrystalline sample by pressing the sample through a special die[6~11]. These experiments mentioned above were usually carried out at ambient or elevated temperatures, but the temperature was still much lower than that for the conventional recrystallization temperature.

Based on Bailey and Hirsh's nucleation mechanism for recrystallization[12], Derby[13~15] developed a theory of dynamic recrystallization and gave a general formula for the grain size and deformation stress. According to the recrystallization kinetics, Hines and Vecchio[4], however, concluded that the observed recrystallized small grains formed in the shock-prestrained copper could hardly be explained in terms of the existed recrystallization kinetics. Based on a "bicrystal" approach, Hines et *al.*[16] used plasticity theory to predict the evolution of subgrain misorientation. They recently proposed a model of progressive subgrain misorientation recrystallization to account for the shear band microstructures. It is apparent that the dislocation activity and plastic deformation enhance subgrain misorientation. There is not as yet an understanding of the relationship between the plastic strain and the subgrain misorientation angle.

The purpose of this paper is to present our recent observations on the recrystallized grains in the shear bands of Monel alloy and a high strength low alloy steel deformed at high strain rate of order $10^5$ s$^{-1}$. The small grains of high purity aluminum heavily deformed by ECAE at room temperature are also observed. Consequently, based on subgrain misorientation, a model of dynamic recrystallization induced by high strain plastic deformation is here further developed. It will be revealed that the dislocation mobility with strain hardening effect during the deformation plays a key role in the recrystallized grain formation.

## 2. Experimental

Three materials used in this study are a high strength low alloy steel, a Monel alloy(Ni-based alloy) and a high purity Al. Their chemical compositions are listed in Table 1. The samples of the steel were quenched from 890°C and then tempered at 600°C for 2 h. An orthogonal cutting test was carried out on a lathe at high cutting speed. Many shear bands

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Table 1 Chemical compositions of three tested materials

| | |
|---|---|
| High strength low alloy steel | C:0.30, Cr:1.03, Mo:0.46, Ni:3.0, V:0.13 |
| Monel alloy | Ni:65.1, Cu:29.5, Al:3.00, Ti:0.48, Fe:0.9, Mn:0.67, C:0.15, Si:0.20 |
| High purity aluminum | Al: 99.999 |

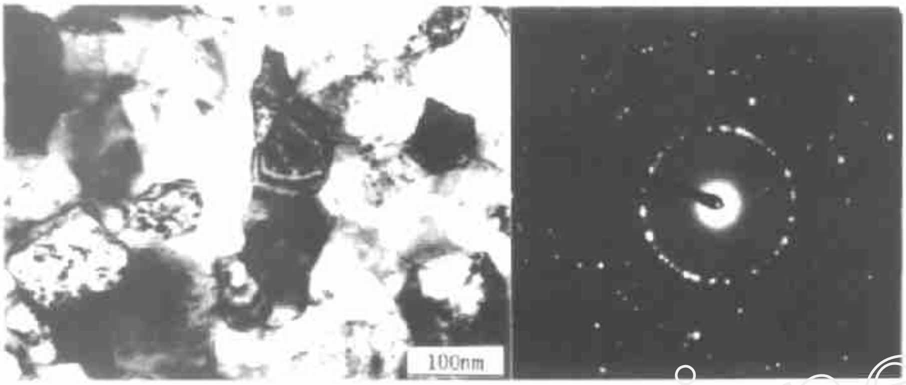


Fig.1 TEM micrographs of recrystallized grains in the shear band formed in the high strength low alloy steel

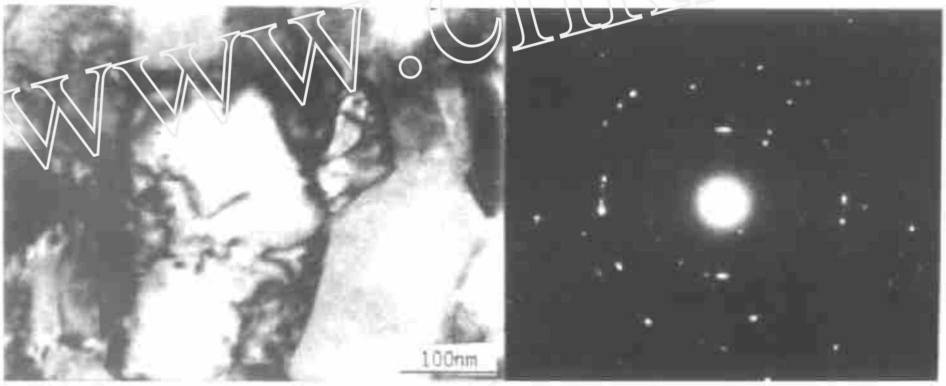


Fig.2 TEM micrographs of recrystallized grains in the shear band in Monel alloy impacted by Hopkinson bar

were observed in the chip. The plastic shear strain ($\gamma$) , strain rate ($\dot{\gamma}$) and shear stress ($\tau$) were evaluated previously, *i.e.* $\gamma$=15.5, $\dot{\gamma}=3.5\times10^5$ $s^{-1}$ and $\tau$=216 MPa[17]. Small recrystallized grains within the bands were observed by TEM as shown in Fig.1

"Hat-shaped" samples of the Monel alloy were impacted by split Hopkinson pressure bar at ambient temperature. The experimental procedure was described elsewhere[18]. The intensely deformed bands with a width of ~20 $\mu$m were produced during dynamic loading. The shear strain, strain rate and shear stress were 10, $4.3\times10^5$ $s^{-1}$ and 350 MPa respectively. TEM micrographs shown in Fig.2 reveal clearly the microstructural characteristics of the recrystallized grains with very sharp grain boundaries in the center of the bands.

Several billets of high purity Al (10 mm× 10 mm×70 mm) with initial grain size of ~1 mm were simply sheared using a new technique of ECAE. The shear strain produced by each passage depends on the channel angle in the die and a strain formulation was given by Iwahashi *et al.*[19]. In our laboratory, a new die with channel angle of 100 degree was designed, and the shear strain obtained by each passage was as high as 1.57. After the billet was pressed through the die for 3 times, the accumulated shear strain amounts to 4.71. Then, a specimen cut from the center of the billet was observed by SEM using electron backscattering channel contrast technique. Figure 3 shows the microcrystalline structure with a grain size of 2~3 $\mu$m, which is quite consistent with the TEM results reported earlier by Nakashima[9] and Iwahashi[10]. The majority of the grain boundaries were sharp, although some of them were still thickly populated by dislocations.

## 3. Kinetics of Grain Formation and Growth for Recrystallization

Cho *et al.*[5] indicated that a lot of cell structures were formed during plastic deformation. They showed that, with increasing strain or strain rate, substantial dislocations were produced and then joined to the cell

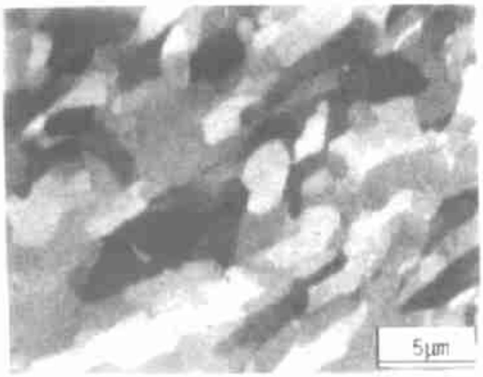


Fig.3 Micrograph of recrystallized grains in high purity Al heavily sheared by ECAE technique

walls, making the subgrain boundaries gradually increasing their misorientation with respect to their neighbors. Adopting this basic idea, the process of dynamic recrystallization is postulated and formulated as follows: During plastic deformation, the complex dislocation structure resulting from high strain rate and/or high strain becomes progressively less uniform. There will be relatively dislocation free cells or subgrains surrounded by dislocation-rich boundaries. These subgrains are separated from each other by peripheries with their respective misorientation. Further deformation induces more dislocations joining to the sub-boundaries with the result that a certain subgrains will become more dislocation free cells but with increasing misorientation with respect to the neighbors. In other words, these subgrain boundaries are becoming movable so that after fulfilling the requirement of a critical size, these cells start to migrate rapidly and the recrystallization begins. So, it is necessary, first, that either a large strain or a high strain rate could progressively concentrate a lot of dislocations to the subgrain boundaries to make them becoming high angle boundaries. Starting from a simple energy balance equation, we obtain:

$$\frac{\mathrm{d}}{\mathrm{d}t}(\eta\pi D^2) = \alpha\mu b^2\pi D^2\delta\frac{\mathrm{d}\rho_{\mathrm{w}}}{\mathrm{d}t} \tag{1}$$

$$\eta = \alpha\mu b\theta(1 - \frac{\ln(\theta)}{4\pi(1-v)\alpha}) \tag{2}$$

where $\eta$ is the specific subgrain boundary energy[20], $D$ is the equivalent diameter of the cell structure (or subgrain), $\alpha$ is a constant of 0.5~1.0, $b$ is Burgers vector, $\delta$ is the cell wall thickness, $\theta$ is the grain boundary misorientation, $\mathrm{d}\rho_{\mathrm{w}}/\mathrm{d}t$ is the dislocation density increasing rate within the cell wall. On the other hand, $\mathrm{d}\rho_{\mathrm{w}}/\mathrm{d}t$ depends on the strain rate, that is[16]

$$\frac{\mathrm{d}\rho_{\mathrm{w}}}{\mathrm{d}t} = \left(\frac{2}{3}\right)^{1/2}\left(\frac{D}{\delta b}\right)\rho_{\mathrm{m}}^{1/2}\frac{\mathrm{d}\gamma}{\mathrm{d}t} \tag{3}$$

where $\rho_{\mathrm{m}}$ is the mobile dislocation density. We still seek the relationship between $\rho_{\mathrm{m}}$ and the strain $\gamma$. This has been given by Gilman[21].

$$\rho_{\mathrm{m}} = (\rho_{\mathrm{o}} + M\gamma)\exp(-\frac{H\gamma}{\tau_{\mathrm{s}}}) \tag{4}$$

where $\rho_{\mathrm{o}}$ is the initial dislocation density, $M$ is a dislocation multiplication coefficient, $H$ is a hardening coefficient, $\tau_{\mathrm{s}}$ is the applied shear stress. During the incubation period when the subgrains are not sufficiently large enough to grow, the continuous deformation could be able to increase their misorientation angles. In order to obtain the changing misorientaion with the increasing strain, we assume that the subgrain size $D$ keeps a stable scale below the critical dimension before growth. So, let $D$ = constant just before grain growth, substituting Eqs.(2), (3) and (4) into Eq.(1), the kinetics for subgrain misorientaion angle ($\theta$) can be described by the following equation,

$$\frac{\mathrm{d}\theta}{\mathrm{d}\gamma} = \frac{4\pi(1-v)\alpha D}{4\pi(1-v)\alpha - 1 - \ln(\theta)}\left(\frac{2}{3}\right)^{1/2} \cdot (\rho_{\mathrm{o}} + M\gamma)^{1/2}\exp(-\frac{H\gamma}{2\tau_{\mathrm{s}}}) \tag{5}$$

This is an expression indicating that the subgrain boundary misorientaion angle is changing with the strain just before recrystallized grain growth. The values of $M$ of some metals were given by Gilman[21] and $H$ could be obtained by fitting the stress *vs* strain curve[21,22]. $\rho_{\mathrm{o}}$ can be evaluated from initial state of the material and $D$ can be determined by microstructural observations of a series of deformed samples. Hines *et al.*[16] gave the stationary subgrain sizes for Cu, Al, Ni and Cr, which were in a range of 100~400 nm. Figures 4(a) and 4(b) show the variations of $\rho_{\mathrm{m}}$ and $\theta$ *vs* the strain $\gamma$ calculated from Eq.(5) by using the appropriate parameters. It can be seen that when $\rho_{\mathrm{m}}$ starts to increase rapidly, the misorientation also show a steep rise until a saturation value is reached. The variation of the misorientation with strain for different metals will be discussed below.

## 4. Discussion

The microstructural evolution in highly deformed zones of metallic materials may be basically separated into following three stages. Firstly, at the beginning of plastic deformation, a great deal of dislocations are produced, the mobile dislocation density is sharply increased [Figs.4(a) and 4(b)] and the dislocation tangles and cell structures start to form. Secondly, with continuous deformation, the cell structures are elongated along shear direction, and the dislocation walls are formed, which results in many small subgrains in the initial grains. Finally, the subgrain boundaries will be progressively populated by moving dislocations, so that they finally become high angle movable boundaries. As soon as the subgrain size reaches a critical dimension which is very small for a heavily deformed metal, rapid recrystallization immediately

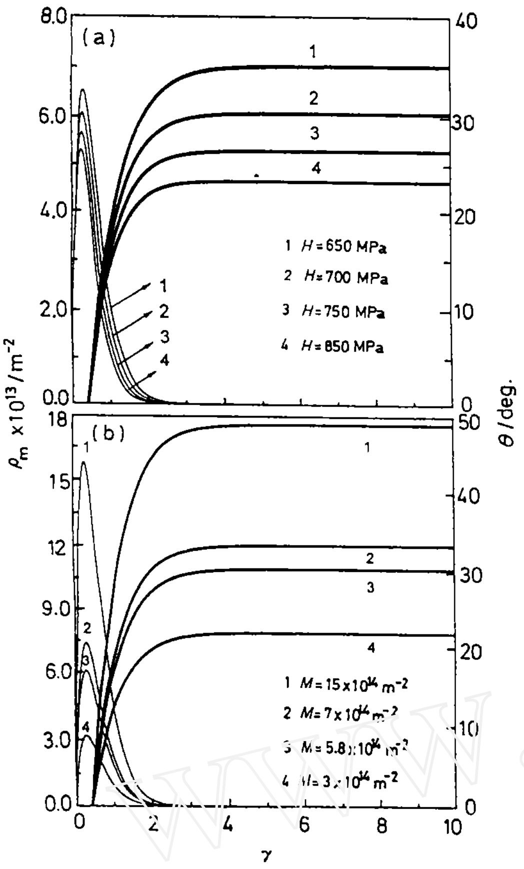


Fig.4 Subgrain misorientation angle ($\theta$) and mobile dislocation density ($\rho_{\rm m}$) *vs* shear plastic strain($\gamma$) curves calculated using different work-hardening coefficient-$H$ (a) and different dislocation multiplication coefficient-$M$ (b), taking $\rho_0=10^{10}{\rm m}^{-2}$ and $\tau_s$=200 MPa as an example, thin lines represent $\rho_{\rm m}-\gamma$, coarse lines represent $\theta-\gamma$

takes place in the matrix. The relationship of the subgrain misorientation angle ($\theta$) with plastic shear strain can be described by Eq.(5). There are two main parameters, $M$ and $H$, which have their different effects on increasing misorientation respectively. Fig.4(a) indicates that the final subgrain misorientation angle decreases with the increasing $H$, which is consistent with the fact that work hardening will hinder grain orientation change. Figure 4(b) shows that the final subgrain misorientation angle is increased with increasing $M$. Therefore, movable subgrain boundaries in Al and Cu are easily formed, so that ultra-fine grain Al and Cu only have been obtained using the ECAE technique.

## 5. Conclusion

Small recrystallized grains were observed within the shear bands of three materials heavily deformed during dynamic loading. The shear bands of Monel alloy was initiated by split Hopkinson pressure bar impact at a shear strain rate of $4.3\times10^5$ s$^{-1}$ and a high strength low alloy steel was cut by the orthogonal cutting machine at a shear strain rate of $3.5\times10^5$ s$^{-1}$. The bulk of high purity Al was heavily deformed by ECAE. The experimental results showed that the high strain and/or high strain rate deformation could induce dynamic recrystallization. Based on dislocation dynamics and subgrain boundarys misorientation change enhanced by plastic deformation, a model for describing the dynamic recrystallization is developed further. The subgrain boundary misorientation angle can be raised to a much higher value if the work hardening coefficient is sufficiently low or the dislocation multiplication coefficient is large enough.

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REFERENCES

[1] H.R.Pak, C.L.Wittman and M.A.Meyers: *in Metallurgical Application of Shock-Wave and High-Strain-Rate Phenomena*, ed. L. E. Murr, K. P. Staudhammer and M. A. Meyers, New York and Basel, 1986, 749.
[2] A.H.Chokshi and M.A.Meyers: *Scripta Metall. Mater.*, 1990, **24**, 605.
[3] U.Andrade and M.A.Meyers: *Acta Metall. Mater.*, 1994, **42**, 3183.
[4] J.A.Hines and K.S.Vecchio: *Acta Metall. Mater.*, 1997, **45**, 635.
[5] K.M.Cho and S.H.Lee: *Acta Metall. Mater.*, 1993, **41**, 923.
[6] R.Z.Valiev and E.V.Kozlov: *Acta Metall. Mater.*, 1994, **42**, 2467.
[7] S.Frerrasse and V.M.Segal: *Metall. Mater. Trans.*, 1997, **28A**, 1047.
[8] J.Wang and Y.Iwahashi: *Acta Metall. Mater.*, 1996, **44**, 2973.
[9] K.Nakashima and Z.Horita: *Acta Metall. Mater.*, 1998, **46**, 1589.
[10] Y.Iwahashi and Z.Horita Z: *Acta Metall. Mater.*, 1997, **45**, 4733.
[11] P.B.Berbon and N.K.Tsenev: *Metall. Mater. Trans.*, 1998, **29A**, 2237.
[12] J.E.Bailey and P.Hirsch: *Proc. Roy. Soc.*, 1962, **267**, 11.
[13] B.Derby and M.F.Ashby: *Scripta Metall. Mater.*, 1987, **21**, 879.
[14] B.Derby: *Acta Metall. Mater.*, 1991, **39**, 955.
[15] B.Derby: *Scripta Metall. Mater.*, 1992, **27**, 1581.
[16] J.A.Hines and K.S.Vecchio: *Metall. Mater. Trans.*, 1998, **29A**, 191.
[17] Q.Li, C.X.Ma and Z.H.Lai: *Mater. Sci. Tech.*, 1996, **12**, 351.
[18] Qiang LI and Yongbo XU: *Acta Matall. Sin.*, 1999, **35**, 49.
[19] Y.Iwahashi and J.Wang: *Scripta Mater.*, 1996, **35**, 143.
[20] W.Shockley and W.T.Read: *Phys. Rev.*, 1949, **75**, 692.
[21] J.J.Gilman: *Appl. Mech. Rev.*, 1968, **21**, 767.
[22] P.P Gillis and J.J.Gilman: *J. Appl. Phys.*, 1965, **36**, 3380.