Dynamic recrystallization behaviour at grain boundaries and triple junctions

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Abstract. Dynamic recrystallization (DRX) behaviour and nucleation mechanisms were investigated using copper and copper alloy bicrystals, tricrystals and polycrystals. New grains were preferentially formed along grain boundaries in the bicrystals. After grain-boundary migration and bulging, nuclei appeared behind the deeply bulged grain boundary regions. The critical strain for nucleation was about one-quarter to one-half of the peak strain. The characteristics of nucleation at a grain boundary depended sensitively on grain boundary character. In copper alloy bicrystals, nucleation was much delayed due to solute drag of migrating grain boundaries. The nucleation at triple junctions, in contrast, took place at a much lower strain. New grain formation at triple junction was stimulated by development of folds. All the new grains were twin-related ($\Sigma 3$) to the matrix and were formed behind the migrating grain boundaries. Therefore, it was revealed that the DRX mechanism in copper and copper alloys was essentially controlled by annealing twin formation. Variant selection of the twinning plane depended sensitively on the direction of the grain-boundary migration and on the geometry, however, was not affected by activated slip plane or dislocation glide. The DRX nucleation mechanisms at grain boundaries and at triple junctions are discussed with respect to grain-boundary migration and annealing twin formation.

1. Introduction
Nucleation of dynamic recrystallization (DRX) is known to occur preferentially near regions of high strain gradient and high dislocation density, e.g., deformation bands [1], particle/matrix interfaces [2, 3], grain boundaries [4, 5], and triple junctions (TJs) [6]. All these are characteristic sites where non-homogeneous deformation takes place. For example, grain boundaries and TJs are locations of inhomogeneous deformation because of the plastic strain compatibility that is necessary to accommodate a dissimilar shape change of the component grains or when grain boundary sliding (GBS) is blocked. This induces preferential nucleation of DRX even at a much lower strain, i.e., at 1/4 to 1/2 of the peak strain. There are a plenty of reports about nucleation mechanisms at grain boundaries. Bailey and Hirsch [7] as well as Luton and Sellars [8] proposed a simple bulging mechanism. This process is known to occur in low-to-medium stacking fault energy metals and alloys. Gottstein et al. [9] proposed that DRX concerns annealing twin formation in low stacking fault energy alloy. Miura et al. [5] have revealed from the experiments using copper bicrystals that nucleation takes place at the largely and deeply bulged areas induced by grain-boundary migration (GBM) and that all the new grains are of annealing twins. The bulging occurs towards the directions where high strain energy and high dislocation density were derived by inhomogeneous plastic deformation. While nucleation at grain
boundaries takes place at rather lower strain region than the peak strain. Nucleation at TJ is even possible at strains below yielding, i.e., without macroscopic plastic deformation. Andiarwanto et al. [10] reported that nucleation at TJs onsets at 1/10 to 1/20 of the peak strain in a Fe-Ni alloy polycrystal. Such nucleation is accelerated and enhanced by the extensive occurrence of GBS. When the sliding is blocked at TJs, the required strain concentrations for GBM that triggers annealing twinning are thus provided. In contrast, TJs play reduced roles during static recrystallization because TJs are thermally stable [11]. TJs can, however, migrate along folds which are formed at TJs by impediment of GBS during high temperature deformation [6]. This enables nucleations at lower strains before macroscopic yielding.

When GBS takes place, grain-boundary steps induced by plastic deformation can impede the sliding. The impediment causes inhomogeneous strain around the bearing steps, which sequentially causes GBM and therefore, formation of new grains. Even while GBS can occur extensively below the yield strain, formation of steps by plastic deformation is required for this mechanism. Therefore, nucleation at grain boundary steps starts at a much higher strain region compare with that at TJs while both are induced by GBS [6,12].

When the most essential DRX mechanism in copper and copper alloys is annealing twin formation, the variant selection would strongly affect the texture formation after DRX. Miura et al. revealed that the twinning-plane variant is selected on the basis of the direction of GBM [5, 12]. In case of nucleation at TJs, however, the twinning-plane selection seemed not definitely controlled by the migration direction of TJs [13]. The mechanisms of the variant selection of twinning plane, therefore, still remain obscure.

In the present paper, DRX and nucleation behaviour in copper and copper alloys are introduced from our basic experiments using bicrystals, tricrystals and polycrystals. And the essential nucleation mechanisms are precisely discussed. The use of bicrystals and tricrystals provided ideal experimental conditions under which the presence of other grain boundaries or TJs could not interfere with the analysis.

2. Experimental

2.1 Procedure of preparation of bicrystal samples and microstructural observations

Orientation controlled bicrystals of 99.99% purity copper containing straight symmetrical [001] tilt boundaries were grown from seed crystals by the Bridgman method. The tilt angles, \( \theta \), were 20° and 64° according to the definition of the symmetrical [001] tilt boundary as described in Fig. 1. Because for the grain boundaries were aligned parallel to the tensile axis, GBS is geometrically difficult except some cases such as GBS for accommodations of the component crystals deformed. For comparison, a \( \theta = 20^\circ \) bicrystal with the boundary inclined 45° against the loading axis was also prepared, in which GBS takes place easily. These bicrystals are referred to as \( \theta = 20^\circ (//) \) and \( \theta = 20^\circ (<) \) bicrystals, respectively. Furthermore, a single crystal, which was cut out of a component single crystal of the \( \theta = 20^\circ (//) \) bicrystal, and polycrystal samples (initial grain size: \( D_0 = 23.3 \mu m \)) were also prepared. The tensile tests were carried out in vacuum at 923 K and at a strain rate of \( 4.2 \times 10^{-4} \text{ s}^{-1} \) on an Instron-type mechanical testing machine. The specimens were deformed to a strain (\( \varepsilon = 0.1 \)) lower than the peak strain and quenched using \( \text{H}_2 \)-gas to freeze the microstructure for metallographic observations. Metallography was
performed on the [001] plane, using optical microscopy after polishing and electrolytic etching. For a crystallographic analysis of the DRXed grains, orientation-imaging microscopy (OIM (Hitachi S-3500/TexSEM Inc.)) was employed.

2.2 Procedure of sample preparation of Cu-Sn-P alloy bicrystals and compression tests

Cu-0.19mass%Sn and Cu-0.66mass%Sn alloy bicrystals having [001] twist 17° grain boundaries were grown by the Bridgman method using seed crystals. After discharge machining to rectangular shaped samples with dimensions of 12 x 9 x 6 mm³, they were deformed in compression in the direction parallel to twist [001] axis, i.e., the direction normal to the grain boundary plane. High temperature compression tests were carried out in vacuum at 1073K and at a strain rate of $\dot{\varepsilon} = 2 \times 10^{-2}$ s⁻¹ to various strains of $\varepsilon = 0.2, 0.5, 1.0$ followed by water quenching. Microstructural observations were carried out on the plane parallel to the compression axis. A polycrystalline Cu-0.66mass%Sn sample with an initial grain size of about 200 µm was also prepared and hot deformed at the same conditions for comparison.

2.3 Procedure of preparation of pure copper tricrystals and compression tests

A 99.9999% high purity copper tricrystal with dimensions of 8 x 8 x 32 mm³ was cut out of an ingot unidirectionally solidified, which was composed of a columnar-grained structure. A schematic illustration of the tricrystal is exhibited in Fig. 2. The details of the four cube compression samples cut out from the large tricrystal as well as the geometries of the grain boundaries with respect to the compression axis are also indicated. A straight TJ line runs through the whole tricrystal. Although the orientations of the component three single crystals remain fixed, the grain boundary geometry changed depending on its position within the tricrystal. The component grains and grain boundaries will be referred to as GI, GII and GIII, and GBI, GBII and GBIII, respectively. GBI, GBII and GBIII are all random type boundaries with misorientation angles of 24.6°, 31.8° and 45.0°, respectively. Such random boundaries are known to slide easily at high temperatures [4]. The cube tricrystal samples were compressed in an Instron-type mechanical testing machine at 673K in a vacuum at a true strain rate of $2.9 \times 10^{-4}$ s⁻¹. After compression tests to a true strain of 0.05, they were water quenched. Microstructural observations were carried out from the direction parallel to the TJ lines.

![Figure 2](image)

**Figure 2** Schematic illustrations of the tested tricrystals cut out from a large columnar polycrystal. The large original tricrystal is composed of a twisted arrangement of grain boundaries. The component single crystals in the small four tricrystals cut out of the large original tricrystal possess completely the same crystallographical orientation [12].

2.4 Procedure of pure copper polycrystal sample preparation and tensile tests

A 99.99% purity polycrystalline copper sheet having an initial grain size of about 460 µm prepared by cold rolling and annealing, was discharge machined to tensile specimens with gauge dimensions of 12
x 4 x 1 mm$^3$. The coarse initial grain size was for easy distinction of newly developed grains. Tensile tests were carried out in vacuum on an Instron-type testing machine at 873 K and various initial strain rates from $4.2 \times 10^{-5}$ to $4.2 \times 10^{-3}$ s$^{-1}$ followed by water quenching. Observations of folds were carried out on the plane normal, and then, samples were mechanically polished and etched for microstructural observations of DRXed grains.

3. Results and discussion

3.1 DRX behaviour in copper bicrystals

3.1.1 Typical flow curves

The true stress vs. true strain curves obtained by tensile tests are shown in Fig. 3. After yielding, both bicrystals show gradual work hardening up to a peak stress followed by work softening due to the extensive occurrence of DRX and, then, eventually the flow stress drops to rupture (Fig. 3 (a)). The different hardening behaviour of the two samples in Fig. 3 (a) is attributed to the orientation geometry in addition to the changes in the activated slip systems. In both geometries, the individual component single crystals ideally deform by non-coplanar double slip. The large work-hardening region compared with polycrystalline materials is, therefore, due to widely appeared stage II deformation, which would be not so sufficient to rapidly achieve critical dislocation density for DRX nucleation. The tensile axis in the $20^\circ$)// bicrystal was close to [016] with an equal Schmid factor in both component crystals. And the tensile axis in the $20^\circ$(<) bicrystal was close to [023] again with the same Schmid factor. This is evident from the work-hardening behaviour which reveals a stage I characteristic in the early stages of deformation for the $20^\circ$(<) sample in contrast to the $20^\circ$)// bicrystal that reveals a stage II soon after yielding. Therefore, the peak strain in the $20^\circ$(<) bicrystal is higher than in the $20^\circ$)// bicrystal but the peak stresses are comparable. When compared with the flow curves of the single crystal, bicrystal and polycrystal, the peak strain appears earliest in the polycrystal sample (Fig. 3 (b)). These peak strains in the bicrystal and single crystal look almost the same and delayed compared with that of the polycrystal. This result indicates that appearance of the peak strain should be strongly affected by the density of grain boundaries and TJs as well as dislocations. This tendency is well known that DRX onsets more easily and earlier with a decreasing grain size [14]. The higher flow stress in the bicrystal than that in the single crystal even while the same component single crystals and tensile direction would be due to

![Figure 3](image_url)

**Figure 3** True stress vs. true strain curves for (a) $\theta = 20^\circ$ bicrystals with boundary parallel (//) to and $45^\circ$ (>) inclined to the loading direction and (b) $\theta = 64^\circ$ bicrystal, the component single crystal and polycrystal. The broken line in (a) indicates the strain of water quenching for structural observations [5, 13].
the formation of a grain-boundary affected zone (GBAZ) to induce larger strain hardening. This will be discussed in detail later.

3.1.2 Evolved microstructure obtained by tensile tests

The surface morphology of the 20° bicrystal (||) after straining to $\varepsilon = 0.1$ is shown in Fig. 4. The conjugated slip traces of the primary and secondary slip planes can be observed. The appeared stage II deformations in Fig. 3, therefore, can be attributed to the conjugated slips. Mild GBM can be also seen to take place in Fig. 4.

The microstructure developed in the $\theta = 20^\circ (||)$, $\theta = 20^\circ (<)$ and $\theta = 64^\circ$ bicrystals at a strain of $\varepsilon = 0.1$ was investigated using optical microscopy (Fig. 5). It is evident that the evolved microstructure differs sensitively depending on the misorientation angle and geometry of the bicrystal. Relatively rough GBM took place when the tensile axis is parallel to the grain boundary (Figs. 5 (a), (c)), while that is quite smooth and uniform in the bicrystal in which extensive GBS was possible (Fig. 5 (b)). In the $\theta = 20^\circ (||)$ bicrystal, the conjugated slip traces of the primary and secondarly, and third and fourth slip planes, respectively. T.D. is the tensile direction [5].

![Figure 4](image1.png)

**Figure 4** Surface morphology formed on the 20° bicrystal (||) after straining to $\varepsilon = 0.1$. A and B indicate the conjugated slip traces of the primary and secondary, and third and fourth slip planes, respectively. T.D. is the tensile direction [5].

![Figure 5](image2.png)

**Figure 5** Evolution of substructure along the grain boundary after straining to $\varepsilon = 0.1$ of bicrystals of (a) $\theta = 20^\circ (||)$, (b) $\theta = 20^\circ (<)$ and (c) $\theta = 64^\circ$. Arrow marks indicate the tensile direction [5,13].
64° bicrystal, new grains were formed along the deeply migrated grain boundary. The strain $\varepsilon = 0.1$ is 2/5 of the peak strain. It is evident, therefore, DRX nucleation preferentially onsets at the grain boundaries. GBS is assumed to promote stress concentration to trigger GBM and DRX nucleation [15, 16]. On the other hand, it is reported that deep GBM, which is related with faster GBM, induces more rapid formation of annealing twins [13]. Figure 5 indicates that the latter should be more effective on the nucleation. In short, grain-boundary serration and GBM should play the most important role for nucleation right from the beginning of deformation.

When slips take place in the component single crystals, GBAZs are soon formed along grain boundaries to accommodate the asymmetrical plastic deformation. It causes complicated plastic deformation as well as accumulation of dislocations along the grain boundary region and, then, should induce deep and severe GBM. Mahajan et al. [17] and Miura et al. [6,13] have reported such annealing twin formation behind migrating grain boundaries. This coincides with the hypothesis proposed by Gleiter that annealing twins are formed by the development of stacking faults during GBM [18]. Such stacking faults should occur more frequently when GBM occurs more rapidly and inhomogeneously as observed in the $\theta = 64^\circ$ bicrystal (Fig. 5 (c)). The twinning process sets off massive DRX along the grain boundary. Thus, the faster GBM causes, the earlier DRX nucleation at grain boundaries takes place. Other influences include the grain-boundary character [19] and the inhomogeneous formation of GBAZs [6, 20] to cause severe serration and deep GBM.

The evolved DRX grains were investigated using OIM and the result is exhibited in Fig. 6. It is clear that all the new grains are of 1st order $\Sigma 3$ annealing twins. It is assumed, therefore, that twinning must be the most essential mechanism of discontinuous DRX. While the new grains were formed behind the deeply migrated grain boundary, the twinning plane is normal to the GBM direction rather than the activated slip plane. This would suggest that twinning-plane variant selection is not related with deformation twinning but with stacking faults during GBM. The mechanism of nucleation and twinning variant selection is illustrated in Fig. 7. Twinning due to staking faults preferentially takes place behind

![Figure 6 OIM map of the 64° bicrystal deformed to $\varepsilon = 0.1$. $\Sigma 3$ boundaries are indicated by white lines. The crystallographical orientation is schematically represented below the OIM map [13].](image)
the deeply and rapidly migrating grain boundary and the twinning-plane variant is selected by the direction of GBM, i.e., the closest normal plane to the GBM direction.

3.2 Effect of element addition on DRX nucleation behaviour

It is revealed from the above experiments using bicrystals that GBM and annealing twin formation should the most essential mechanism for DRX nucleation in pure copper. However, a question of what can be the mechanisms for the nucleation when GBM is suppressed would arise. To reveal the DRX behaviour and the mechanisms of nucleation in copper alloys in which GBM is suppressed by the solute drag effect, Cu-Sn alloy bicrystals were grown and tested in compression at 1073 K. The obtained flow curves are shown in Fig. 8. Although deformation was carried out at a sufficiently high temperature, the deeply and rapidly migrating grain boundary and the twinning-plane variant is selected by the direction of GBM, i.e., the closest normal plane to the GBM direction.

Figure 7 Schematic illustration representing the nucleation mechanism and twin-variant selection. All the straight dotted lines are twin boundaries. The tetrahedron at the top illustrates the crystallographic orientation of the upper grain.

Figure 8 True stress vs. true strain curves of Cu-0.19 mass%Sn and Cu-0.66 mass%Sn alloy bicrystals with [001] twist 17° boundaries. The compression axis was parallel to [001].
peak stress appeared only in Cu-0.19mass%Sn alloy bicrystal at around $\varepsilon = 0.8$. Work hardening due to stage II deformation is still continued even at a strain of 1.0 in the Cu-0.66mass%Sn alloy bicrystal. In both samples, the peak strain became much larger compared with that in pure copper (Fig. 3). This result clearly indicates that the onset of DRX is notably delayed by Sn addition, even though the effect of strain rate is considered.

The change in the microstructure during compression tests was investigated using OIM. The summarized results are displayed in Fig. 9. In the Cu-0.66mass%Sn alloy bicrystal, new grain formation associated with an annealing twin was strongly suppressed. Instead of twins, substructure and low angle boundaries are developed along grain boundaries and they become more evident with increasing strain. That is, new grains were formed along a grain boundary accompanied by gradual increase in misorientation. This gradual increment in the misorientation should be induced by formation of GBAZ rather than the mechanisms of continuous DRX. Contrary to that, in the Cu-0.19mass%Sn alloy, new grains associated with twins are formed at $\varepsilon = 0.5$ along a grain boundary and it spreads into the whole grain interior at $\varepsilon = 1.0$. The strain where extensive DRX is achieved is much larger compared with that in pure copper. However, the microstructure developed at $\varepsilon = 0.5$ (Fig. 9 (e)) should be noticed that a few twins were formed. Most of the new grains were mainly composed of sub and low angle boundaries. It is particular to see that the substructure was significantly developed in the grain interiors. The shape and the geometry of the substructure look closely related with slip lines in the component grains. Therefore, it is assumed that the substructure and GBAZ were developed at low to medium strain regions, and then, extensive GBM took place accompanied by annealing twin formation at a high strain region due to highly stored strain energy and dislocation density in the Cu-0.19mass%Sn alloy (see Fig. 9 (e)). Any way, in both bicrystals, a mechanism similar to continuous DRX can also possibly play an important role in grain refinement especially in the grain interiors. The microstructure of Cu-0.66mass%Sn alloy polycrystal deformed to $\varepsilon = 1.0$ is shown in Fig. 10. Plenty of DRX grains associated with annealing twins were formed along the initial grain boundaries. This result indicates that DRX is

![Figure 9 OIM maps of Cu-0.66mass%Sn, (a), (b), (c), and Cu-0.19mass%Sn, (d), (e), (f), alloy bicrystals with [001] twist 17°grain boundaries tested in compression at 1023 K and at a strain rate $\dot{\varepsilon} = 2 \times 10^{-2}$ s$^{-1}$ to various strains. White, thin black, bold black and brown lines indicate sub, low angle, high angle and twin boundaries, respectively.](image-url)
not itself difficult in Cu-0.66mass%Sn alloy and is assisted by presence of a high density of grain boundaries and TJs. This view is in good agreement with the report that the onset of DRX becomes easier with decreasing grain size [14] and with the result of much delayed DRX nucleation in the bicrystals (Fig. 9). The role of TJs on DRX nucleation will be discussed in the next session.

3.3 DRX nucleation at TJs
Even while TJs are one of the most important components in polycrystalline materials, studies about the effect of TJs on DRX nucleation are still not so popular. Observed results of DRX nucleation at TJs in tricrystals and polycrystals are introduced below.

3.3.1. Nucleation at TJs in tricrystals
Typical OIM maps of the grains nucleated at the TJs in tricrystals at a strain of $\varepsilon = 0.05$, which is shortly after yielding, are displayed in Fig. 11. Because of the large Schmid factors roughly between 0.49 and
0.47 in the component grains, double slips began to take place soon after the yielding. It is evident in Fig. 11 that DRX grains have most preferentially nucleated at TJs. It should be noted that all DRX grains were accompanied by deep migration grain boundaries and TJs. In total 13 nuclei were observed in whole tricrystal samples. In the case of Fig. 11 (b), TJ migrated towards GI and G II and a twin was formed in G III behind the migrating TJ. Therefore, the most important DRX mechanism at TJs in copper appears also the formation of an annealing twin. It is interesting to see that TJs can easily migrate during high temperature deformation. Such behaviour is in contrast with the high thermal stability of TJs during static annealing [11]. The observed migration of TJs is brought by the assistance of GBS that takes place and formation of folds during high temperature deformation [6].

Miura et al. have shown from the experiments using copper tricrystals that TJs are the most preferred sites for nucleation under conditions where extensive GBS takes place [12]. In such a case, DRX nucleation began to appear only at around $\varepsilon = 0.01$ strain, which is about 1/10 to 1/20 of the peak strain. This was explained in terms of the sequential occurrence of GBS, folding in the GBS blocking grain, strain-induced migration of a TJ along the fold and twinning behind the migrating TJ. Though GBS is sensitively influenced by the geometry and grain-boundary character [21], the geometry independent nucleation in the present tricrystal should be attributed to the random type grain boundaries that can thus slide easily. A few DRX grains were nucleated also at grain boundaries (Fig. 11(a)), while this was always accompanied by nucleation at TJs.

A typical photograph of folding is displayed in Fig. 12. The fold was formed at a TJ because of blocking of GBS. The direction of the fold is not related with slip planes in the blocking grain. The fold is, however, more readily formed when the directions of GBS and of slip in the grain impeding GBS are close [6]. In Fig. 12, TJs and grain boundaries have already slightly migrated which can be found from the slip traces evolved on the surface.

3.3.2. Nucleation at TJs in polycrystals

Pure copper polycrystal samples were tensile tested up to a strain $\varepsilon = 0.06$ at maximum, which was 2/3 and 1/3 of the peak strains when tested at $4.2 \times 10^{-3}$ to $4.2 \times 10^{-3}$ s$^{-1}$ respectively. The change in the probability of nucleated TJs and the number of formed folds are summarized as a function of strain in Fig. 13. It is marvelous that DRX nucleation at TJs could be recognized even at the strain of $\varepsilon = 0.01$ which was at around yielding. And the amount of DRX nucleation at TJs increased almost linearly with strain. Furthermore, the probability increased as the strain rate decreased. Because GBS takes place more easily and more extensively with a decreasing strain rate [16], the deformation concentration and therefore folding at TJs becomes more easily provided. This tendency is clearly shown by the observation exhibited in Fig. 13 (b), in which the number of the detected folding increased with strain. However, the probability of nucleation at TJs reduced to half when the TJs were composed of twin (Σ3) boundaries. This would be because the twin boundary hardly slides [21]. Hence, the deformation
concentration and folding at TJs are difficult. Actually, nucleation at TJs appears less significant under the conditions where GBS is difficult [13].

Figure 14 shows the analysed results of crystallographical orientation distribution of the nucleated DRX grains at TJs. Here, only the smallest value or the smallest misorientation angle among the orientation relationships between the mother grains and the new DRX grains were plotted. It is revealed from Fig. 14 that more than 80% of the nucleated DRX grains at TJs were $\Sigma 3$ twins. This result is quite similar to those in copper tricrystals tested in compression [22, 23]. These results suggest that the most essential and operative DRX mechanism at TJs in copper, in other words, metallic materials of low-medium stacking fault energy, is twinning. The result that the nucleated grains excepting $\Sigma 3$ orientation had quite large misorientations would also well explain the nature of discontinuous DRX behaviour in copper.

![Figure 13](image_url)

**Figure 13** Summarized results of changes in (a) probability of DRX nucleation and (b) number of folds formed at triple junctions with increasing strain.

![Figure 14](image_url)

**Figure 14** Strain-rate dependence of misorientation and $\Sigma$ value distributions of nucleated DRX grains at triple junctions.
A mechanism of DRX nucleation at TJ is schematically drawn in Fig. 15. Folding takes place in the grain blocking the extensive GBS. Grain boundaries migrate along the fold. A twin is formed behind the fastest migrating grain boundary. Because TJ can migrate fastest, twins are most preferentially formed at the TJ. GBS towards the TJ becomes more notable due to softening after nucleation at the TJ, which would accelerate the TJ migration. This process can operate even before yielding.

It is known that DRX becomes easier with a decreasing grain size [14]. It is assumed that a high density of grain boundaries, which are the one of the most preferential nucleation sites, stimulates DRX. Serration and bulging necessary for the nucleation mechanism at grain boundaries becomes, however, more difficult with a decreasing the grain size due to the limited space. Nucleation at TJs, therefore, becomes more important when the grain size becomes finer. Nucleation at quadruple junctions must be most operative in a fine-grained structure in such conditions.

4. Summary
The nucleation of dynamic recrystallization (DRX) at grain boundaries and triple junctions (TJs) were systematically investigated using copper and copper alloy bicrystals, tricrystals and polycrystals. DRX nucleation started at a grain boundary at strains much lower than the peak strain. This proves that nucleation takes place preferentially at a grain boundary. However, nucleation at TJs can onset in a further a lower strain region around yielding. All these nucleations were always accompanied by migration of grain boundaries and TJs along a grain-boundary affected zone or fold. Because grain-boundary sliding causes stress concentration and the formation of fold, DRX is much affected by its occurrence. All nucleated DRX grains were annealing twins to one of the adjoining grains in the pure copper bicrystals and the tricrystals, which substantiates the idea that the DRX nucleation mechanism and thus the origin of discontinuous DRX in metals of low stacking fault energy is annealing twin formation. Annealing twins appeared behind the migrating boundaries, preferentially at locations that are the most deeply bulged. However, annealing twin formation becomes difficult by element addition to retard grain boundary migration. In this case, mechanisms of continuous DRX can contribute to nucleation. The twinning-plane variant of the nucleated twins behind the migrating grain boundary was not selected at random but was related to the direction of grain boundary migration. Activated slip planes or dislocation motion did not affect this valiant selection.

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References
[1] Blaz L, Sakai T and Jonas JJ 1983 Metal Sci. 17 609.
[2] Humphreys FJ 1979 Acta Mater. 27 1801.
[3] Ferry M and Humphreys FJ 1996 Acta Mater. 44 1293.
[4] Miura H, Aoyama H and Sakai T 1994 J. Japan Inst. Metals 58 267.
[5] Miura H, Sakai T, Mogawa R and Gottstein G 2004 Scripta Mater. 51 671.
[6] Miura H, Sakai T, Andiarwanto S and Jonas JJ 2005 Phil Mag. 85 2653.
[7] Bailey JE and Hirsch PB 1962 Proc. Roy. Soc. (London) A267 11.
[8] Luton MJ and Sellars CM 1969 Acta Metall. 17 1033.
[9] Gottstein G, Zabardjadi D and Mecking H 1979 Metal Sci. 13 223.
[10] Andiarwanto S, Miura H and Sakai T 2004 Tetsu-To-Hagane 90 21.
[11] Gottstein G, Ma Y and Schvindlerman LS 2005 Acta Mater. 53 1535.
[12] Miura H, Sakai T and Jonas JJ 2006 Scripta Mater. 55 167.
[13] Miura H, Sakai T, Mogawa R and Jonas JJ 2007 Phil. Mag. 87 4197.
[14] Sakai T and Jonas JJ 1984 Acta Metall. 32 189.
[15] Miura H, Aoyama H and Sakai T 1994 J. Japan Inst. Metals 58 267.
[16] Miura H, Ozama M, Mogawa R and Sakai T 2003 Scripta Mater. 48 1501.
[17] Mahajan S, Pande SC, Iman AM and Rath BB 1977 Acta Mater. 48 2633.
[18] Gleiter H 1969 Acta Met. 17 1421.
[19] Winning M, Gottstein G and Shvindlerman LS 2001 Acta Mater. 49 211.
[20] Peralta P and Laird C 1997 Acta Mater. 45 3029.
[21] Monzen R, Sumi Y, Kitagawa K and Mori T 1990 Acta Metall Mater. 38 2553.
[22] Miura H, Andiarwanto S and Sakai T 2003 Mater. Sci. Forum 426-432 4387.
[23] Miura H, Andiarwanto S and Sakai T 2002 Mater. Trans. 43 494.