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Recovery by triple junction motion in heavily deformed metals

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Abstract. A number of processes may occur during recovery, which reduce the stored energy and coarsen the structure, paving the way for the nucleation of recrystallization. In this review, recovery is discussed based on the initial deformed structure. Recovery in samples deformed to low-to-medium strains is briefly considered, whereas recovery in samples deformed to much higher strains is discussed in more detail. An example is Al, where a new and important recovery mechanism is triple junction motion, which can remove thin lamellae and consequently increase the thickness of neighbouring lamellae. This recovery mechanism therefore coarsens the structure and causes a gradual transition from a lamellar to a near equiaxed structure preceding recrystallization. This mechanism is discussed in conjunction with structural pinning, which stabilizes the coarsening microstructure, thereby balancing the forces controlling the rate of recovery by triple junction motion. Operation of triple junction motion in other heavily deformed metals, such as Ni and Cu, is also briefly discussed.

1. Introduction

Recovery is traditionally defined as all annealing processes occurring in deformed materials without migration of high angle boundaries [1], whose misorientation angles are typically larger than 15°, and it is usually considered as a precursor of recrystallization [2]. Recently, we have documented that coarsening by triple junction motion is an important restoration mechanism in heavily deformed Al subdivided by lamellar boundaries, containing a substantial fraction of high angle boundaries [3]. In that work, triple junction motion has been classified as a recovery mechanism since it occurs before primary recrystallization, although migration of high angle boundaries is heavily involved.

The current paper is focused on new findings on recovery of heavily deformed Al, including mechanisms, kinetics, structural pinning and effects on nucleation of recrystallization. Both deformation structures at different strain levels and recovery behaviours at low-to-medium strains are briefly considered as background. This background is based on polycrystalline metals of medium-to-high stacking fault energies, processed under conditions where plastic deformation takes place by glide of dislocations. Finally, recovery in heavily deformed metals other than Al is also discussed.

2. Deformation microstructure

Deformation of metals of medium-to-high stacking fault energies (e.g. Al, Ni, Cu and Fe) to a low strain typically produces cell-block structures, where a cell block is a group of cells where the same sets of slip systems operate [4]. Most of the dislocations are stored in cell boundaries and cell-block
boundaries, whereas regions between these boundaries have a relatively low dislocation density. The structural morphology changes as the strain increases. After deformation to a high strain, for example by cold rolling, the structure is characterized by extended lamellae evolved from cell blocks [5,6]. The lamellar boundaries are finite, forming triple junctions where they meet, and there are interconnecting boundaries linking lamellar boundaries. The two types of boundaries and three types of triple junctions are illustrated in figure 1a in the longitudinal section, containing the rolling direction (RD) and the normal direction (ND), and in figure 1b in three dimensions. Each Y-junction is formed by three lamellar boundaries, each pair of H-junctions is formed by two lamellar boundaries and an interconnecting boundary between them, and each r-junction is formed by three interconnecting boundaries [3].

Figure 1. (a) An illustration of a lamellar boundary structure in the longitudinal section, where the lamellar boundary (red) spacing is $D$, the interconnecting boundary (cyan) spacing is $l$, and the lamella length is $L$. (b) An illustration of the three-dimensional boundary structure of the marked region in (a). Lamellar boundaries are shown in different shades of red, with the top layer transparent, and interconnecting boundaries are shown in cyan. One Y-junction is shown in orange; all H-junctions are shown in blue; and all r-junctions are shown in green. The TD is the transverse direction [3].

Both cell-block boundaries at low strains and lamellar boundaries at high strains are geometrically necessary boundaries (GNBs), whereas both cell boundaries at low strains and interconnecting boundaries at high strains are incidental dislocation boundaries (IDBs) since they are assumed to form by mutual trapping of glide dislocations [7,8]. For both GNBs and IDBs, the average boundary spacing decreases and the average misorientation angle across the boundary increases when the strain is increased. However, the rate of change is very different for the two types of boundaries, indicating
different mechanisms controlling their formation and evolution. In parallel, the density of three types of triple junctions all increases with increasing strain at different rates [3].

In addition to the change in the structural morphology and structural parameters, the stored energy of deformation also changes when the strain is increased. It increases monotonically with increasing strain over a large strain range although at a decreasing rate [9,10].

3. Recovery in metals deformed to low-to-medium strains
The recovery of metals deformed to low-to-medium strains has been studied extensively more than half a century ago, as reviewed by Bever [11] and Perryman [12]. The early recovery studies were primarily focused on the release of the stored energy and the changes in the physical and mechanical properties, and recovery processes were discussed in terms of annihilation of excess vacancies, mutual annihilation of dislocations with opposite Burgers vectors and rearrangement of dislocations into low angle subgrain boundaries [11,13,14]. Later experimental observations by transmission electron microscopy (TEM) and theoretical analyses showed that subgrain growth is also an important recovery process, either by subgrain coalescence (e.g. [15,16]) or by subgrain boundary migration (e.g. [17,18]). While dislocation annihilation and reconfiguration might be the dominating recovery mechanism in metals of low stacking fault energies, subgrain growth by subgrain boundary migration seemed to be more important in metals of high stacking fault energies (e.g. [19-21]). The growth of subgrains is of particular interest since large subgrains partly or entirely surrounded by high angle boundaries may develop into nuclei for recrystallization. Figure 2 shows an example, where the spacing of boundary dislocations in a low angle twist boundary (A-B) connected to a high angle boundary (C-D) is not uniform. The spacing near the high angle boundary is larger than that far away from this boundary. This is considered as evidence of dislocation absorption in a high angle boundary and a precursor of subgrain coalescence, which may lead to nucleation of recrystallization [22]. For reviews in recovery, see also [23] and [24].

Figure 2. A bright-field micrograph of a low angle twist boundary (A-B) which is linked to a high angle boundary (C-D). The dislocation spacing within the net of the low angle boundary is found to increase from about 27 nm near A to about 33 nm near B [22].
Specific recovery mechanisms have been analysed based on kinetic equations where parameters are dislocation density, subgrain size and flow stress [16,20,25-28]. However, recovery consists of a number of complex processes, involving a number of parameters, and in order to describe the global recovery kinetics, empirical rate equations [11] are often used.

If P is a material property, decreasing during annealing, then a logarithmic decay of the property corresponds to the following rate equation:

$$\frac{dP}{dt} = -\frac{c_1}{t},$$

where \(c_1\) is a constant and \(t\) is the annealing time. However, the relationship between the property and the time is indirect, and equation (1) leads to an infinite rate at the beginning of recovery. To overcome the above two problems, the first-order reaction rate equation may be used:

$$\frac{dP}{dt} = -c_2 P \exp\left(-\frac{Q_0}{RT}\right),$$

where \(c_2\) is a constant, \(Q_0\) is the activation energy independent of temperature, \(R\) is the gas constant, and \(T\) is the absolute temperature. Equation (2) assumes a constant activation energy. However, it has been shown that the apparent activation energy for recovery increases as recovery proceeds (e.g. [20,29,30]), and therefore the recovery rate equation first proposed by Kuhlmann [31] (later as Kuhlmann-Wilsdorf) may give a better description:

$$\frac{dP}{dt} = -c_3 \exp\left(-\frac{Q_0 - \beta P}{RT}\right),$$

where \(c_3\) and \(\beta\) are constants. However, equation (3) has a problem at the end of recovery, i.e. the rate never goes to zero. A more flexible equation has been proposed by Borelius and co-workers [32]:

$$\frac{dP}{dt} = -c_4 P \exp\left(-\frac{Q_0 - \beta' P}{RT}\right),$$

where \(c_4\) and \(\beta'\) are constant. Equation (4) may be recognized as a first-order reaction rate equation with an apparent activation energy depending on the extent of recovery which already has taken place. It may describe the whole range of recovery as suggested recently [30,33].

4. Recovery in Al deformed to high strains

In recent years there has been an increasing interest in metals and alloys deformed to large strains with a very high strength. The structure is on a fine scale and the stored energy is high. Consequently their thermal stability is of significant interest. Al has a high stacking fault energy, so cross slip is easy and recovery plays an important role during annealing. For highly strained Al, more than half of the stored energy can be released by recovery processes before the first recrystallization nucleus is observed [30]. Due to its large extent of recovery, Al has been favoured in many studies (e.g. [20,34,35]) and some detailed work even demonstrated the dependence of the recovery behaviour on local texture [34,35]. During plastic deformation excess vacancies are produced, but they annihilate rapidly below room temperature in Al almost independent of the strain and the impurity level [36,37]. Therefore in this section Al is chosen to demonstrate recovery after deformation to high strains, and recovery is discussed in terms of dislocations, interconnecting boundaries and lamellar boundaries, whereas excess vacancies are ignored.

4.1. Initial stage of recovery

In a recent study of recovery in commercial purity aluminium AA1050 (99.5% purity) cold rolled to a true strain of 5.5 (99.6% thickness reduction), the initial stage of recovery was investigated by isothermal annealing in the temperature range 5-100 °C [38]. Post-mortem microstructural
characterization by electron backscatter diffraction (EBSD) and TEM showed that the spacings of lamellar boundaries and interconnecting boundaries were almost unchanged during annealing, pointing to a negligible effect of subgrain coalescence and migration of low and high angle boundaries. However, the density of dislocations in regions between these boundaries was found to decrease significantly during annealing, indicating dislocation annihilation as an important recovery mechanism in the initial stage of recovery. In addition, extensive TEM observations of thin foils at the same areas before and after annealing not only confirmed annihilation of loose dislocations and dislocation bundles during low temperature annealing, but also revealed sharpening of interconnecting boundaries, indicating reorganization and possibly annihilation of dislocations in the boundary. An example of such TEM observations is shown in figure 3.

Figure 3. Weak-beam dark-field TEM micrographs showing annealing of dislocations and reorganization of interconnecting boundaries during storage of a TEM thin foil at room temperature for 9 days. (a) Before storage; (b) after storage.

In the same study [38], the recovery kinetics was followed by hardness tests. Due to the small recovery window examined, a constant activation energy was assumed and thus equation (2) was applied. The analysis led to a small apparent activation energy, 60-86 kJ/mol. However, this apparent activation energy may not directly correspond to an ideal diffusion mechanism since different diffusion species may be involved, different diffusion paths may contribute and ideal diffusion mechanisms may be affected by the deformation structure. Nevertheless, the low apparent activation energy is consistent with annihilation of dislocations between boundaries and within interconnecting boundaries, and the energy barriers for diffusion processes may be lowered by the presence of defects, e.g. a high density of dislocations and finely spaced boundaries.

4.2. Recovery by triple junction motion
While the initial stage of recovery only had a limited influence on the structure and the strength of the deformed Al, the following stage of recovery led to both significant coarsening of the microstructure and a marked decrease in strength [33]. In a recent study of recovery in heavily cold-rolled commercial purity aluminium AA1050 (true strains 2, 4 and 5.5), thermally activated migration of
triple junctions formed by lamellar boundaries (i.e. Y-junctions) was identified as a key recovery mechanism, leading to removal of thin lamellae and uniform coarsening of lamellar structures [3].

During annealing of deformed samples in the temperature range 120-220 °C, general post-mortem TEM and EBSD investigations showed that the lamellar boundary spacing measured along the ND was increased but the lamellar morphology was largely maintained. After longer annealing periods, a gradual decrease of the structural aspect ratio led to a gradual transition from the lamellar morphology into a near equiaxed one [3]. Extensive TEM observations were carried out for a number of thin foils at the same areas before and after annealing. It was found that the process responsible for the removal of lamellar boundaries (and thus the increase of lamellar boundary spacing) was Y-junction motion. One example is shown in figure 4 where the thin foil was annealed at 120 °C. The Y-junction migrated up, removing the middle lamella in black contrast and increasing the lamellar boundary spacing of two neighbouring lamellae while keeping the lamellar morphology, in agreement with the general post-mortem observations. At higher annealing temperatures, Y-junction motion was found to be much more frequent, and an example is shown in figure 5 for annealing at 220 °C. Significant Y-junction motion took place in an initially finely spaced region, leading to doubling of the lamellar boundary spacing locally after annealing for 1 h. The microstructure after annealing evolved into being more equiaxed, but it kept a clear lamellar morphology.

Figure 4. An example of Y-junction motion observed in the longitudinal section of Al cold rolled to a true strain of 5.5 and annealed at 120 °C for different total periods of time [3].
Figure 5. TEM observations of Y-junction motion during annealing at 220 °C. (a) Microstructure of the same area before and after annealing for 1, 8, and 60 minutes. (b) Sketches of the same area at different states corresponding to (a). For the states after annealing, each sketch combines the current state with the previous state to show the structural evolution. Solid thin lines represent lamellar boundaries which were unchanged; dashed lines represent boundaries which had migrated with new positions represented by solid bold lines. The areas swept by migrating boundaries are shown in grey, and the directions of the Y-junction motion are indicated by arrows. Lamella “A” was completely removed after 8 minutes and lamella “B” coarsened at the beginning due to removal of lamella “A” but then was shortened greatly by Y-junction motion [3].
Besides observations by TEM on thin foils, Y-junction motion was also observed by EBSD on the surface of bulk samples deformed to different strains [3]. Figure 6 shows an example, where the differences in the microstructure of the same area at different cumulative annealing periods are indicated. As shown in this case, Y-junction motion still played an important role although the lamellar boundary spacing had increased significantly (from 0.24 to 0.57 µm) after annealing for 2 h. Further coarsening by Y-junction motion and general boundary migration led to nucleation of recrystallization, which was observed after annealing for 96 h at this temperature [39].

Figure 6. Examples of Y-junction motion in the longitudinal section of Al deformed to a strain of 5.5 and annealed at 220 °C. (a) Annealed for 2 hours; (b) annealed for 4 hours accumulatively. The colours are coded according to the orientation of each pixel after mild noise reduction using Channel 5. High angle boundaries (>15°) are shown in black and low angle boundaries (1.5−15°) in white. (c) A sketch of the lamellar boundaries combining the two states, following the same markings as in figure 5. The marked region is magnified in (d) [40].

Based on observations and analyses, Y-junction motion was proposed as a dominant recovery mechanism in a finely spaced lamellar structure [3]. The sketch shown in figure 7 illustrates how such a structure coarsens via Y-junction motion and evolves into a more equiaxed structure gradually, and after further annealing some coarsened subgrains may develop into nuclei of recrystallization.
Figure 7. Schematic diagrams illustrating structural coarsening via Y-junction motion. (a) The original deformation structure with typical lamellar boundaries (bold black lines) and interconnecting boundaries (thin grey lines). This lamellar structure coarsens via Y-junction motion with arrows showing the directions of motion. (b)-(d) Intermediate coarsening steps. (e) A more equiaxed structure has evolved after coarsening via Y-junction motion. (f) A modified equiaxed structure considering also the drag of interconnecting boundaries at H-junctions (i.e. interconnecting boundaries resist widening) [3].

4.3. Triple junction motion and structural parameters

The geometry of a Y-junction is important in determining its stability, in particular the spacing $D$ and the dihedral angle $2\theta$ associated with the middle lamella. The dihedral angle and the lamellar boundary spacing shown in figure 8 were measured from TEM micrographs of deformed samples, and the categorization into two groups was based on the annealing results – circles for not-migrated Y-junctions and dots for migrated ones. The plot shows that: (i) the estimated dihedral angle appeared to increase with increasing lamella spacing although the exact value depended on the local configuration (e.g. affected by neighbouring H-junctions); (ii) a migrating Y-junction generally had a small dihedral angle and the corresponding shortening lamella generally had a small spacing; and (iii) neither a lamella spacing below ~50 nm nor a dihedral angle below ~30° was observed. However, the large dispersion of data shown in figure 8 indicates that other factors may also play a role in determining the stability of a Y-junction [3].
Figure 8. Distribution of the dihedral angle and the lamella spacing of Y-junctions in Al deformed to a strain of 5.5. The pairs of dihedral angle and lamella spacing were estimated from TEM micrographs taken from the longitudinal section. Open circles represent random measurements, whereas solid points represent measurements of Y-junctions observed to have migrated after annealing at 120 °C for 5 hours [3].

Most of the migrating Y-junctions were composed of three high angle boundaries or a combination of high angle boundaries and low-to-medium angle boundaries, but cases of migrating Y-junctions formed by three low-to-medium angle boundaries were also found although much less frequently. The deformation structures were subdivided into typical rolling texture components, and Y-junction motion was examined for each rolling texture component but no significant orientation preference was found. Y-junction motion was more frequently found at higher temperatures during recovery annealing, and also in samples deformed to higher strains, which have smaller boundary spacings and higher stored energy [3].

The driving force for Y-junction motion comes from many sources, e.g. Y-junction line energy and strain energy from dislocations, but the principal source was considered to be from the boundary energy (grain boundary surface tension). However current theories of boundary mobility [41,42] cannot explain the rapid Y-junction motion observed at low annealing temperatures, nor the absence of dihedral angles below 30° or lamellae thinner than 50 nm. This has led to the tentative suggestions that (i) the triple junction mobility may depend on its dihedral angle, being very large for a sharp triple junction, and (ii) an internal stress may build up at sharp triple junctions during deformation, leading to mechanically assisted triple junction motion and removal of very thin lamellae [3].

4.4. Activation energy
The coarsening kinetics of the same heavily cold-rolled Al was studied recently [39]. In that study, deformed Al was subjected to a series of isothermal annealing treatments in the temperature range 140-220 °C. Both EBSD and electron channel contrast (ECC) imaging were used to characterize the microstructural evolution, and the measured lamellar boundary spacings are shown in figure 9 for samples that had undergone microstructural coarsening before recrystallization. The increase of the lamellar boundary spacing approximately followed a linear relationship in the log-log plot and the slope \(1/n\) increased with increasing annealing temperature. Based on equation (2), the annealing time \(t\) and the annealing temperature \(T\) (K) required for achieving the same degree of recovery were related by the following equation:

\[
\ln t = c + \frac{Q}{RT},
\]

where \(c\) is a constant. Since the activation energy for recovery increases significantly in the course of recovery (e.g. [20,29,30]), equation (5) may only apply to each small recovery window, where the
variation of the activation energy is small. Based on equation (5), the activation energy for each small recovery window was determined from $(\ln t)$ vs $1/T$ plots. The analysis led to a low activation energy (110 kJ/mol) at the beginning of coarsening and a high one (240 kJ/mol) at the end of uniform coarsening, i.e. when the lamellar boundary spacing had increased to 0.9 µm (figure 10). In this interval, the microstructural coarsening was dominated by Y-junction motion.

![Graph showing spacing (µm) vs. time (h) for different temperatures](image)

**Figure 9.** EBSD (solid symbols) and ECC (open symbols) data of the average lamellar boundary spacing during recovery of cold-rolled Al. The spacing was measured by the line intercept method along the ND. Different slopes ($1/n$) of linear least squares fittings in log–log plots are indicated for different temperatures [39].

![Graph showing variation of apparent activation energy](image)

**Figure 10.** Variation of the apparent activation energy during microstructural coarsening of cold-rolled Al [39].
The estimated apparent activation energy at the beginning of coarsening was close to that for dislocation annihilation in the initial stage of recovery, and the activation energy at the end of uniform coarsening was close to that for the subsequent recrystallization, which was previously reported to be 230-235 kJ/mol for AA1050 [43,44]. The agreement in these numbers suggests a continuous increase of the apparent activation energy during annealing, including both recovery and recrystallization.

It should also be noted that, based on the results shown in figures 9 and 10, the microstructural coarsening of the same material at other annealing temperatures may be predicted by a simple extrapolation. For example, an extrapolation to 300 °C predicts that about 1 minute is needed for the lamellar boundary spacing to increase to 0.9 µm, in good agreement with experimental results [45]. It was indeed reported [45,46] that Y-junction motion still played an important role in Al during annealing at 300 °C, where the recrystallization kinetics was accelerated, and that recovery by Y-junction motion affected boundary populations in recovered microstructures and thus subsequent recrystallization.

In another study [40], the recovery kinetics was studied based on hardness measurements. Instead of being divided into small recovery ranges, the whole kinetics was fitted to equation (4), which assumes that the apparent activation energy increases linearly with decreasing stored energy. The results were generally consistent with the analysis based on boundary spacings presented above.

4.5. Structural pinning

In a recent study, Y-junction motion in heavily cold-rolled Al was studied by in situ TEM [47]. The in situ observations showed that the migrating Y-junctions were pinned by encountered interconnecting and lamellar boundaries, which slowed down the recovery process and led to a stop-go migration pattern. The dynamic pinning force for Y-junction motion was directly observed by the bowing of encountered boundaries and their component dislocations as illustrated in figures 11 and 12. This pinning mechanism stabilizes the recovered microstructure, i.e. the structure is stabilized by balancing the driving and pinning forces controlling the rate of Y-junction motion.

Figure 11. Schematic drawing that shows the pinning and depinning interactions between a moving Y-junction and interconnecting boundaries (dashed lines), including the bowing and slight extension of the interconnecting boundary during annealing. The arrow indicates the direction of Y-junction motion, D is the thickness of the shortening lamella and φ is the misorientation angle of the interconnecting boundary [47].
It was also found in TEM foils cut from bulk annealed samples that Y-junctions were typically in the vicinity of attached interconnecting or lamellar boundaries (see figures 11b and 12b). These attached boundaries increased the dihedral angles at the Y-junctions, thereby reducing the driving force for Y-junction motion and stabilizing the partially recovered structure. Quantitative measurement showed that the average projected distance from a Y-junction to the nearest attached position was only 94 nm, which was about 1/3 of the distance for a random distribution of Y-junctions [47]. Such a big difference indicated that Y-junctions were not randomly distributed, but formed stable configurations, in agreement with in situ observations. Therefore, the effect of these deformation-induced boundaries is paradoxical: on one hand they store deformation energy and provide the driving force for recovery and recrystallization, but on the other hand they stabilize the microstructure by retarding or pinning Y-junction motion and boundary migration.

5. Recovery in other metals deformed to high strains
In other metals deformed by dislocation glide, e.g. Ni and Cu, significant structural differences exist, which may affect the recovery mechanisms and kinetics. For example, both the spacings between lamellar boundaries and the dihedral angles are significantly smaller in Ni and Cu, and as a consequence recovery by triple junction motion may be facilitated in these metals. Besides, the stored energy in Ni and Cu is higher than that in Al in the high strain region. However, there are also factors against recovery in Ni and Cu, for example the competition between recovery and recrystallization. In Al there is a significant recovery range with a large decrease in mechanical properties before the onset of recrystallization. On the other hand Ni and Cu usually show only a small decrease in hardness before recrystallization starts, and structural changes related to recovery processes are expected to be small. Based on the above points, it is difficult to speculate whether recovery by triple junction is more or less pronounced in Ni and Cu, which also represent a lower stacking fault energy and a higher melting point. Preliminary experiments have been carried out on high purity Ni and Cu, showing less pronounced triple junction motion. This observation suggests that other factors have to be taken into account, for example structural pinning, which based on structural parameters may be different in other metals. This is to be explored in future research together with changes in other parameters.

Three important factors are the stacking fault energy, solutes and particles. It is well known that the stacking fault energy plays a key role in the recovery of dislocation networks, especially in pure metals, and therefore compared to Al smaller stacking fault energies in Ni and Cu lead to less recovery of dislocation networks, which may provide a stronger structural pinning effect. Solutes are also very important since they can strongly interact with dislocations, impede dislocation motion and thereby increase structural pinning. The mechanisms associated with solute dislocation interactions are diverse and include elastic size effects, as well as changes to shear modulus and stacking fault energy. Solutes may also segregate at boundaries, resulting in higher activation energies for boundary migration [20,39] and thereby delay recrystallization [48]. This delay gives a larger recovery window for triple
junction motion, which has been shown to operate at lower activation energies (figure 10) and may be less affected by solutes due to its special geometry, i.e. solutes may be deposited at the tailing boundary. Moreover, when there is a dispersion of small particles, Zener drag will decrease the driving pressure for recrystallization and thereby delaying recrystallization, whereas triple junction motion may be less affected due to its high driving force when the lamellar boundary spacing is fine and the dihedral angle is small.

Therefore, future research may be carried out to extend the window of recovery by delaying recrystallization kinetics through alloying [48,49] or by replacing cold deformation with warm deformation. This is to reduce the driving pressure for recrystallization while maintaining a large fraction of deformation induced high angle boundaries at high strains. Based on such structures, it will also be possible to optimize annealing treatments of metals and alloys deformed to large strains to balance their mechanical properties, e.g. strength and ductility.

6. Concluding remarks
The recovery of heavily deformed metals have been demonstrated with a focus on Al, where triple junction motion has been identified as an important mechanism, resulting in structural coarsening and therefore being a precursor of recrystallization. In situ observations showed that migration of Y-junctions was a stop-go migration process, where a rate reduction was caused by the interaction of migrating junctions with deformation induced boundaries. This pinning mechanism stabilized the structure by balancing the driving forces, thereby increasing the thermal stability of the highly strained material. This is of importance when developing ultra-strong metals and alloys by plastic deformation to very high strains.

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