Chapter 8

RECRYSTALLIZATION OF ORDERED MATERIALS

8.1 INTRODUCTION

There is significant interest in the use of ordered intermetallic compounds as high temperature structural materials. However, use of these materials is frequently limited by their poor toughness. Thermomechanical processing may be used not only to form the materials, but also to break-up the cast structure, reduce inhomogeneity, refine the grain size and optimise microstructure and texture (Morris and Morris-Munoz 2000). For superplastic applications (Nieh and Wadsworth 1997), control of grain size during processing and service is critical, and in some cases, control of grain shape and recrystallized state may also be important (e.g. McKamey and Pierce 1992, Zhang et al. 2000).

Few of the potentially useful intermetallics can be extensively deformed at room temperature, and there is most interest in hot deformation and either static or dynamic recrystallization. In this chapter we make some comparisons of deformation, recovery, recrystallization and grain growth in ordered alloys with the behaviour of conventional metals. Although there has been extensive research into thermomechanical processing of ordered materials, much of the work is on complex multi-phase materials such as the TiAl-based alloys (e.g. Yamaguchi 1999) in which the main interest is in optimising the phase distributions, and we will not consider these materials. Even for the simpler single-phase intermetallics, there are many unanswered questions regarding the annealing behaviour, as discussed in the review by Baker (2000).
8.2 ORDERED STRUCTURES

8.2.1 Nature and stability

The two ordered structures (or superlattices) which have been most extensively studied with respect to the annealing behaviour are the $L1_2$ structure associated with $A_3B$ alloys such as Cu$_3$Au, Ni$_3$Al, Ni$_3$Fe etc., and the $B2$ structure typical of $AB$ compounds such as CuZn, FeCo, NiAl etc. The $L1_2$ structure is face centred cubic with the $A$ atoms located on the face centred sites (1/2, 1/2, 0 etc.) while the $B2$ structure is body centred cubic with the $B$ atoms at the body centred site (1/2, 1/2, 1/2). The $B2$ structure is subject to a further significant ordering process in which a $DO_3$ structure is produced. The unit cell consists of eight $B2$ unit cells arranged as a $2 \times 2 \times 2$ cube with alternate cells having $A$ and $B$ atoms at the cell centre. Examples of this structure include the important materials based on Fe$_3$Al.

In some materials the ordered structure is absent above a critical ordering temperature ($T_c$), but in others the ordered state is stable up to the melting point. In many respects materials of the latter type, which are called permanently ordered materials, are closely related to chemical compounds. Details of $T_c$ for a number of significant materials are given in table 8.1.

In materials that are capable of being disordered, the retention of ordering during heating, or its development during cooling, is measured by the ordering parameter ($S$). $S$ takes values between 1, corresponding to perfect ordering and 0, corresponding to the fully disordered state. However, the extent of ordering below $T_c$ does not follow the same pattern in all materials. In some cases, e.g. FeCo, $S$ decreases slowly as the alloy is heated, and the disordered state can be retained at room temperature by quenching. In other cases, e.g. Cu$_3$Au, the disordered state cannot be retained on quenching and there is an abrupt change to high values of $S$ at temperatures just below $T_c$. The two types of behaviour are shown in figure 8.1.

In materials that can be disordered, the structure below $T_c$ is characterised by volumes called domains, where there is a high degree of order. In a fully ordered material, neighbouring domains, in which the $A$ and $B$ atoms have different positions, are separated by a surface called an antiphase boundary (APB) and it is found in practice that such boundaries follow simple crystallographic planes in the parent lattice. APBs are of critical importance to the deformation and annealing behaviour of ordered materials, because they occur not only from the impingement, during ordering, of

<table>
<thead>
<tr>
<th>Material</th>
<th>$T_c$ (°C)</th>
<th>Material</th>
<th>$T_c$ (°C)</th>
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<tbody>
<tr>
<td>Cu$_3$Au</td>
<td>390</td>
<td>Fe$_3$Al</td>
<td>~540</td>
</tr>
<tr>
<td>Ni$_3$Fe</td>
<td>500</td>
<td>(Co$<em>{78}$Fe$</em>{22}$)$_3$ V</td>
<td>910</td>
</tr>
<tr>
<td>CuZn</td>
<td>454</td>
<td>Ni$_3$Al</td>
<td>1638 (M.Pt.)</td>
</tr>
<tr>
<td>FeCo</td>
<td>~725</td>
<td></td>
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</tr>
</tbody>
</table>
growing domains but also from the passage of dislocations (see below). In partly disordered materials the domains are separated from each other by disordered regions.

8.2.2 Deformation of ordered materials

A perfect dislocation in the disordered lattice is merely a partial dislocation in the ordered structure and its movement in the superlattice leaves behind a plane that is an APB. In B2 structures for example, the expected dislocation has a Burgers vector of $a/2 \langle 111 \rangle$ and its passage leads to the creation of an APB. In order to avoid the energy increase associated with the production of such a boundary, the moving dislocations in an ordered structure are coupled in pairs (super-dislocations) so that the second dislocation restores the ordered state. These two dislocations are linked by a small area of antiphase boundary.

The details of the slip processes in ordered B2 alloys have been discussed by Baker and Munroe (1990, 1997). The commonly observed dislocations have Burgers vectors corresponding to $\langle 100 \rangle$, $\langle 110 \rangle$ and $\langle 111 \rangle$ but only the first and last of these are found at low temperatures. In many alloys (e.g. CuZn), there is a change from $\langle 111 \rangle$ slip to $\langle 100 \rangle$ slip at temperatures near 0.4–0.5 Tm. However, in others (e.g. FeAl), the temperature at which the transition to $\langle 100 \rangle$ slip occurs is also governed by the composition. There is at present, no clear explanation for the change from $\langle 111 \rangle$ to $\langle 100 \rangle$ slip in these materials. For $\langle 111 \rangle$ dislocations in B2 superlattices, the slip planes are {011}, {112} and {123} but whether or not the coupled $\langle 111 \rangle$ partials can cross slip, depends on the energy of the APB that links them. For large APB energies the partials are close together and recombination and cross slip are easy. If the energy is very low, cross slip of the individual partials becomes possible. For intermediate values cross slip is more difficult and pile ups are to be expected.

A detailed study of the dislocations present in the $L1_2$ alloy Al$_{64}$Ti$_{28}$Fe$_8$ deformed in compression, has been made by Lerf and Morris (1991). At room temperature, undissociated $\langle 110 \rangle$ dislocations move through the lattice, but $\langle 100 \rangle$ dislocations...
remain pinned. At 500°C, many dislocations dissociate into pairs of a/2 < 110> partials, thereby constituting mobile super-dislocations on {111} planes. At 700°C, the super-dislocations are distributed evenly between the {001} and {111} planes. Cross slip between octahedral planes is increasingly prevalent at temperatures > 500°C.

The very limited low temperature ductility of many materials which remain ordered up to the melting point, is attributable in part to the retention of order and hence to the difficulty of dislocation movement, and also to grain boundary embrittlement in some intermetallics. However, some ordered alloys, such as boron-doped Ni₃Al, can be extensively deformed at room temperature (e.g. Ball and Gottstein 1993a, Yang and Baker 1997), and others can be deformed by techniques such as ball milling (Jang and Koch 1990, Koch 1991, Gialanella et al. 1992) or equal channel angular extrusion (ECAE) Semiatin et al. (1995).

It is often found that a recrystallized metal, particularly one with a small grain size, is more ductile than the cold-worked material. However, there are some examples of intermetallics that are more ductile when only recovered or partially recrystallized (Baker 2000). For example, McKamey and Pierce (1992) examined the room temperature strength and ductility of an Fe₃Al-based alloy using specimens that had been heat treated to relieve stresses or produce controlled amounts of recrystallization. Maximum ductility (8% elongation) was associated with the stress relieved (recovered) state, and in contrast the fully recrystallized material showed only 3.5% elongation.

During extensive plastic deformation, many of the super-dislocations become uncoupled or dissociated, and the material becomes progressively disordered (Koch 1991, Ball and Gottstein 1993a). An investigation by Dadras and Morris (1993) of a Fe₆₈Al₂₈Cr₄ alloy which had a DO₃ structure found that after milling, about 75% of the dislocations present were uncoupled and single.

The stored energy after deformation is seen from equation 2.6 to be proportional to the dislocation density (ρ) and the square of the Burgers vector (b). Values of b are different for disordered and ordered materials, and in L₁₂ compounds, b = a/2 < 110> in the disordered state, but after ordering b doubles to a < 110>. If the dislocation density is similar, we therefore expect the ordered material to have a larger stored energy than the disordered alloy, and there is some evidence for this (Baker 2000). However, the situation is complicated, because as discussed above, an ordered material may disorder during plastic deformation. A higher stored energy will tend to accelerate recrystallization or lower the recrystallization temperature (§7.2), and there is some evidence that an alloy which is ordered before deformation recrystallizes at a lower temperature than the same alloy disordered before deformation (Clareborough 1950, Baker 2000).

8.2.3 Microstructures and deformation textures

Although there has been much discussion of the mechanism of dislocation movement in ordered materials, there is surprisingly little known about the nature of the microstructures and textures developed during thermomechanical processing, as may be seen from the review of Yamaguchi (1999).
Boron-doped alloys based on Ni$_3$Al can be rolled at room temperature, and have been extensively studied by Gottstein and colleagues. Ball and Gottstein (1993a) rolled a boron-doped Ni$_{76}$Al$_{24}$ compound which has an L1$_2$ structure, to >90% reduction at room temperature. They found that at intermediate strains the microstructure consisted of clusters of microbands together with shear bands (fig. 8.2); the latter were of the intersecting type normally associated with alloys of low stacking fault energy. No evidence of cell formation was observed at any level of strain. At strains below 0.4, very thin (~0.05 µm) microbands were observed, and these authors also reported that at higher strains, somewhat thinner shear bands, classified by them as copper type shear bands, appeared between the brass type shear bands formed earlier.

The deformation texture of this material after rolling to 92% reduction is shown in figure 8.3. It was described as a weak copper-type texture consisting of the components, Brass, $\{110\}<112>$ and Copper, $\{112\}<111>$ and with a broad spread between the Goss, $\{110\}<001>$ and S, $\{123\}<634>$ orientations. This rolling texture is unexpected, because shear bands of the type shown in figure 8.2 which are common in materials with low-medium stacking fault energy, are normally associated with strong brass-type textures at this level of reduction (§3.6.4). Ball and Gottstein detected no evidence of deformation twinning, although these have since been reported by Chowdhury et al. (1998), who also found evidence of a structural change to a DO$_{22}$ structure after large strains. The development of a ‘copper type’ texture (§3.2.1) in a microstructure resembling that of a low stacking fault energy fcc material (§2.5) is unusual and there is a need for further investigations.

Kawahara (1983) examined an FeCo-2V alloy; this material which has a B2 structure can be retained in the disordered condition by quenching. Extensive shear banding occurred in both ordered and disordered specimens after rolling. The first shear bands developed after only 10% reduction, and by 50% reduction the specimens displayed microstructures similar to that shown in figure 8.2, in which the shear bands were profuse and had the characteristics of the bands normally found in fcc metals with low stacking fault energy. The rolling texture of a specimen rolled to 70% reduction was said
to consist of \{001\}<110>, \{112\}<110>, \{111\}<110> and \{111\}<211> components and to be similar to that of iron-based materials.

The development of microstructures and textures during warm rolling of B2 ordered polycrystalline alloys based on Fe3Al has been investigated. Huang and Froyen (2002) found that the grains elongated during deformation and that the boundaries remained straight. The dislocation structures were found to be reasonably random, with no evidence of cell formation. Most dislocations were found to be of Burgers vector \( <111> \). Raabe (1996) reported textures near \{111\}<110> on warm rolling similar material, and showed the experimental textures to be in good agreement with Taylor calculations based on the operation of \{110\}<111> and \{112\}<111> slip systems. Morris and Gunther (1996) have shown that the deformation textures in Fe3Al are affected by the alloy composition, the processing temperature and the degree of order.

8.3 RECOVERY AND RECRYSTALLIZATION OF ORDERED MATERIALS

The recovery and recrystallization of ordered structures is complicated by the fact that several quite different processes may be occurring simultaneously in the material, viz., the restoration of the ordered state and those associated with recovery and recrystallization. The kinetics of these processes are quite different and the behaviour
is complex. Reviews of the literature by Cahn (1990) and Cahn et al. (1991) summarize the observations as follows:

- There is a drastic reduction in grain boundary mobility in the presence of atomic order, which severely retards recrystallization.
- A temperature range may exist where recrystallization does not occur even though it occurs at lower and higher temperatures.
- The pre-recrystallization behaviour depends on the composition of the alloy and the annealing temperature. The deformed alloy may soften, harden, or remain unchanged.

In the following account the L1₂ and B2 structures are considered separately.

### 8.3.1 L1₂ structures

#### 8.3.1.1 Recovery

In a significant early study Roessler et al. (1963) examined Cu₃Au after rolling to 63% reduction (ε ~ 1). Both ordered and previously disordered material were used. It was found (fig. 8.4) that in both cases there was a hardness increase when annealing was carried out below the ordering temperature of 390°C; the increase was greatest for material deformed in the disordered state. The maximum values were associated with an increase in S from 0.50 to 0.85. It is well known that there is a relationship between hardness and domain size, and Stoloff and Davies (1964) have shown that the hardness is a maximum at a critical small domain size. The strain–age-hardening shown in figure 8.4 is attributable to rapid domain growth as order returns.

![Fig. 8.4. Microhardness of rolled Cu₃Au as a function of annealing time at 288°C, (Roessler et al. 1963).](image-url)
The alloy Ni$_3$Fe, which has T$_c$ ~ 500°C, was found by Vidoz et al. (1963) to harden during annealing in much the same way as Cu$_3$Au. The effect was greatest at the stoichiometric composition and fell off rapidly as the composition deviated from ideality; off-stoichiometric alloys with > 32% Fe or > 78% Ni softened by recrystallization. In addition to the hardness increase, the stoichiometric alloy showed a strongly enhanced work hardening rate after annealing; this change is believed to be due to dislocation drag involving APB tubes.

Considerable attention has been paid to the family of L1$_2$ alloys based on the composition (FeCoNi)$_3$V (Liu 1984). Cahn et al. (1991) rolled the alloy, (Co$_{78}$Fe$_{22}$)$_3$V, in both the ordered and disordered conditions to reductions up to 50% ($\varepsilon \approx 0.8$) and annealed above and below T$_c$ (910°C). In this alloy S remains high up to T$_c$ and recrystallization was strongly retarded with a retardation factor of ~300X just below T$_c$. In the case of initially ordered material, both hardness and tensile measurements showed substantial softening prior to, and during the early stages of recrystallization as shown in figure 8.5a. However, initially disordered material showed substantial hardening before and during the early stages of recrystallization as shown in figure 8.5b. The increase in hardness, during recovery, of the material deformed in the disordered state (fig. 8.5b) is attributable to hardening induced by ordering during the annealing. The decrease of hardness of the initially ordered material (fig. 8.5a) is due to normal dislocation recovery processes. The hardness decrease of (Co$_{78}$Fe$_{22}$)$_3$V shown in figure 8.5a is in contrast to the increase of hardness during annealing of the initially ordered Cu$_3$Au shown in figure 8.4. It is thought that the latter behaviour is a consequence of the larger rolling deformation resulting in a substantial amount of disordering, which leads to order hardening on annealing.

Fig. 8.5. Hardness changes during the annealing of deformed (Co$_{78}$Fe$_{22}$)$_3$V. (a) Initially fully ordered material rolled to 25% reduction (● = microhardness, × = macrohardness); (b) Initially disordered material rolled to 50% reduction, (Cahn et al. 1991).
We conclude that during recovery of deformed weakly-ordered alloys at temperatures below $T_c$, there are two processes occurring.

- **Normal dislocation recovery**, which results in softening, and
- **Re-ordering**, if the material was disordered before or during deformation, which results in hardening.

The net effect on the hardness during a recovery anneal will therefore depend on the amount of deformation, the rate of re-ordering and the rate of dislocation recovery.

A strongly ordered alloy such as Ni$_3$Al tends to soften on recovery (Baker 2000). This may be because such a material disorders less extensively during deformation than does a weakly-ordered alloy.

### 8.3.1.2 Recrystallization

A second classic study of Cu$_3$Au, concerned in this case with recrystallization, was that of Hutchinson et al. (1973). The disordered alloy was rolled to 90% reduction ($\varepsilon \sim 2.5$) and isothermally annealed (fig. 8.6). The kinetics of ordering were such that order was fully restored at 330–380°C, and the retardation of recrystallization in the presence of order is clearly seen. Hutchinson et al. established that the retardation could not be attributed to inhibition of nucleation in the presence of order and that the significant factor was inhibition of grain boundary migration. This was believed to be due to the increased diffusion distances associated with boundary migration through an ordered lattice. The recrystallization textures were also determined; for annealing above $T_c$ a very weak texture was found, but a strong copper-type deformation-type texture developed at temperatures below $T_c$. By comparing their results with those for pure copper and pure gold, Hutchinson et al. estimated that at $T_c$ the retardation factor due to the presence of a superlattice is $\sim 100 \times$. Cahn and Westmacott (1990) have carried out similar experiments with the alloy (Co$_{78}$Fe$_{22}$)$_3$V and find an even greater retardation factor of $300 \times$.

![Image](image.png)

**Fig. 8.6.** Fraction recrystallized vs annealing time for Cu$_3$Au reduced 90% by rolling, (Hutchinson et al. 1973).
The annealing behaviour of the boron-doped Ni$_3$Al alloy referred to in §8.2.3 has been investigated by Gottstein et al. (1989), Ball and Gottstein (1993b) and Escher et al. (1998). In the initial work, details of the recrystallization kinetics were determined and recrystallization was reported to be slow. The JMAK exponent was 2.2 and the activation energy for recrystallization was found to be 110 kJ/mol which was less than half that found by Baker et al. (1984). In view of the high value of stored energy, which provides a large driving force for recrystallization, Gottstein et al. (1989) attributed the sluggish recrystallization behaviour to a reduced grain boundary mobility in the presence of ordered structures.

Before recrystallization, subgrain formation was found to occur preferentially near the grain boundaries, and recrystallization was found to initiate by strain induced boundary migration at low strain levels. After larger strains, nucleation was associated with shear bands (fig. 8.2) and microbands and equiaxed grains with diameter 0.2 $\mu$m developed at these inhomogeneities. Annealing twins were common in the new grains. In the volumes between the bands, nucleation was much slower and the grains that eventually developed were around three times larger. This situation is directly comparable to the case of low-zinc brass reported by Carmichael et al. (1982) (fig. 7.16 and §7.4.1).

In polycrystals of this material, the recrystallization texture was reported to be almost random, and Ball and Gottstein (1993b) attributed this to the locally high nucleation rate in the shear bands and microbands and the very low grain boundary mobility of ordered structures. Chowdhury et al. (2000) confirmed the weak recrystallization texture and related it to a reversion of the DO$_{22}$ structure formed during deformation to the stable L1$_2$ structure and to the influence on the recrystallization, of twins formed in the later stages of deformation.

8.3.2 B2 structures

FeCo: The annealing behaviour of B2 alloys is different in many respects to that just described, and amongst such materials FeCo has been extensively studied. This alloy has a critical temperature of $\sim$725°C, and the disordered state can be retained by quenching. On deformation, FeCo workhardens at similar rates in both the ordered and disordered states. The kinetics of re-ordering in FeCo deformed in the disordered state are retarded (Stoloff and Davies 1964) in contrast to the behaviour of alloys such as the L1$_2$ phase Cu$_3$Au in which acceleration is observed.

The recrystallization behaviour of FeCo containing small amounts of V or Cr has been studied by Buckley (1979) and Rajkovic and Buckley (1981). If cold worked material is annealed below Tc then interaction between the ordering and recrystallization reactions leads to the very complex pattern of behaviour summarised below.

- Above $\sim$725°C (Tc) recrystallization is rapid.
- Between 600°C and 725°C the alloy orders rapidly and this is followed by slow recrystallization.
- Between 475° and 600°C, homogeneous ordering occurs at a moderate rate, dislocation recovery takes place but no recrystallization occurs. It is presumed that the driving force for recrystallization is insufficient for grain boundary migration through the superlattice.
Between 250° and 475° the ordering reaction is slow and partial recrystallization is possible. Ordering is more rapid in the vicinity of the migrating grain boundaries.

**Fe₃Al:** On recovery, there is extensive recovery of the tangled dislocation structures formed on hot working, into well organised low angle boundaries (Morris and Lebouef 1994, Huang and Froyen 2002), accompanied by a decrease in hardness, and an increase in the degree of long-range order. Recrystallization originates at heterogeneities such as deformation bands (§7.6.4) and by SIBM (§7.6.2) as shown in figure 8.7.

**β-brass:** A detailed study of the annealing of β-brass was carried out by Morris and Morris (1991). As pointed out earlier, the amount of order decreases continuously in this phase as the temperature rises to, and through \( T_c \), and the annealing behaviour (fig. 8.8) does not show the marked discontinuity seen in figure 8.6 for Cu₃Au. Electron microscopy showed that recovery of the dislocation structure was rapid at temperatures well below \( T_c \), but decreased as \( T_c \) was approached and this was attributed to short-range ordering. The increasing rate of recrystallization as \( T_c \) is approached is to be expected, and the activation energy of 145 kJ/mol for boundary migration in this range is close to that for diffusion of copper in the ordered structure. The decreasing rate of recrystallization with increasing temperature above \( T_c \) is however, surprising.

### 8.3.3 Domain structures

An important difference between permanently ordered alloys and those which undergo an order/disorder transformation is the presence of an antiphase domain structure in recrystallized material.

Weakly-ordered compounds such as Cu₃Au and FeCo show profuse APBs after recrystallization (e.g. Hutchinson et al. 1973, Cahn 1990, Cahn and Westmacott 1990,
Yang et al. (1996). However, strongly-ordered compounds such as Ni₃Al show few if any. Yang et al. (1996) suggested that this difference is because grain boundaries in weakly-ordered compounds are partially disordered, and that APBs are created within this disordered zone as the boundary migrates. The APBs will then act as a drag on the boundary, thus explaining why the boundaries, although disordered, move rather slowly compared to the fully disordered material. However, in strongly-ordered compounds, most of the boundaries are ordered and migration is hindered by the presence of long range order right up to the boundary plane, necessitating the two atomic species migrating to their correct sites as the boundary moves, with no consequent formation of APBs.

8.4 GRAIN GROWTH

Most of the studies of grain growth in ordered alloys report isothermal grain growth kinetics which are similar to those of disordered alloys, with the usual relationship between grain size and time as given by equation 11.7. Studies of grain growth in FeCo containing 2%V, have shown that the kinetics are described by this relationship with the exponent $n \sim 2$. If however, the rate constant is plotted against 1/T the results shown in figure 8.9 are observed (Davies and Stoloff, 1966). A normal Arrhenius relation exists above $T_c$ but at lower temperatures the plot is curved and there is a discontinuous change in the rate constant at $T_c$, showing that even a small degree of order significantly slows grain growth. Huang and Fryen (2002) have reported a similar decrease in the grain growth rate in Fe₃Al when the ordering is increased.

Ball and Gottstein (1993b) investigated grain growth after recrystallization in the boron-doped Ni₃Al discussed in §8.3.1, and in the temperature range 800–1150°C. They found that the kinetics were of the form of equation 11.7, with a growth exponent $n$ which
varied somewhat with temperature, having a mean value of 3. The activation energy for grain growth was found to be 298 kJ/mol, close to that for diffusion of Ni in Ni$_3$Al.

Baker (2000) has discussed the effect of stoichiometry on the rate of grain growth in ordered Fe–Co. Figure 8.10 shows the effect of stoichiometry on grain growth in Fe–Al, and it is seen that the rate of grain growth increases as the alloy deviates from

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**Fig. 8.9.** The grain growth rate constant in FeCo–V as a function of temperature, (Davies and Stoloff 1966).

**Fig. 8.10.** Grain growth kinetics for various Fe–Al alloys at 1200°C, showing the effect of deviation from stoichiometry, (after Schmidt et al. 1989).
stoichiometry. As iron substitutes for aluminium, the alloy becomes more disordered and the rate of diffusion increases.

The effect of interstitial solutes such as boron, on grain growth rates is complex. In Ni$_3$Al, it is found that boron has little effect on boundary mobility (e.g. Cahn 1990, Yang and Baker 1996). However, boron has a very strong inhibiting effect on grain growth in FeAl (Yang and Baker 1996). This may be due to the nature of the interstitial sites in the two lattices.

8.5 DYNAMIC RECRYSTALLIZATION

Dynamic recrystallization is frequently observed during the hot deformation of ordered alloys. It appears to be rather similar to the process which occurs in metals, and is to be expected in materials where recovery is slow and there is a retention of large amounts of stored energy (§13.3).

Evidence of grain boundary migration and dynamic recrystallization has been obtained by Baker and Gaydosh (1987), and Imayev et al. (1995) in specimens of FeAl. At temperatures in the range 600–700°C, where a <111> superdislocations operate (Baker and Munroe 1997), dynamic recrystallization is very slow. However, at temperatures in the range 750–900°C, where a <100> dislocations operate, dynamic recrystallization proceeds rapidly. This difference in behaviour has been ascribed to the ability of the <100> dislocations to climb more rapidly (Imayev et al. 1995), although the effects of changes in the grain boundary mobility with temperature should not be discounted.

Aretz et al. (1992) and Ponge and Gottstein (1998) have investigated dynamic recrystallization during the hot compression of large-grained boron-doped Ni$_3$Al. The stress–strain curve exhibited a peak, typical of metals undergoing dynamic recrystallization (§13.3), and dynamic recrystallization occurred at the old grain boundaries, with the new grains forming the typical ‘necklace microstructure’ described in §13.3.3. The mechanism of dynamic recrystallization appears to involve progressive lattice rotations in the grain boundary regions, and may be similar to that found for magnesium (Ion et al. 1982) and some minerals, which is discussed in §13.4.2. In the early stages of dynamic recrystallization, local boundary migration was observed and the orientations of the first layer of grains in the necklace were rather close to those of the original grains. However, as the structure developed with increasing strain, the orientations became more random, and this was attributed to the formation of annealing twins, and grain rotation arising from superplastic-type deformation in the dynamically recrystallized regions.

8.6 SUMMARY

The recovery, recrystallization and grain growth behaviour of ordered alloys shows many general similarities to disordered metals. Although the large number of material...
and processing variables means that much remains to be clarified, the following conclusions can be drawn (Cahn 1990, Baker 2000):

- If deformed in the ordered state, ordered compounds recrystallize more rapidly than if deformed in the disordered state, and this is attributable to the larger stored energy.

- Grain boundary mobility is severely reduced by ordering, thus retarding both recrystallization and grain growth. Deviations from stoichiometry generally increase diffusivity and increase boundary mobility.

- During recovery of a weakly ordered alloy the hardness will usually decrease if $T > T_c$ and increase if $T < T_c$. Permanently ordered alloys tend to soften on recovery.

- In some weakly-ordered alloys, a temperature range may exist where recrystallization does not occur, even though it occurs at higher or lower temperatures.

- Weakly ordered alloys contain APBs after recrystallization, whereas strongly ordered alloys do not.