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Recrystallisation Mechanism and Annealing Texture in Aluminium-Copper Alloys

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The correlation between the mechanism of recrystallisation and the annealing texture of aluminium-copper alloys was investigated by transmission electron microscopy and selected area diffraction, and pole figure determination by X-rays. Continuous recrystallisation by sub-grain growth leads to preservation of the rolling texture, while recrystallisation by motion of a high-angle boundary produces a cube texture as in pure aluminium. The conditions under which the different modes of recrystallisation occur and the reasons for the formation of the two types of textures are discussed on the basis of microscopic mechanisms.

1. Introduction

Many commercial alloys such as steels and aluminium alloys are usually not used as homogeneous solid solutions but as phase mixtures. The recrystallisation behaviour of such alloys is known to be difficult to predict. It depends on whether second phase particles are precipitated before plastic deformation or during recrystallisation, and on the nature and the dispersion of the second phase. According to a recent review of this subject [1], healing out of dislocations can take place by two basically different mechanisms:

(i) Formation of sub-boundaries, and continuous growth of sub-grains with increasing misorientation, until angles greater than those between sub-grains (i.e. high-angle boundaries) are developed.

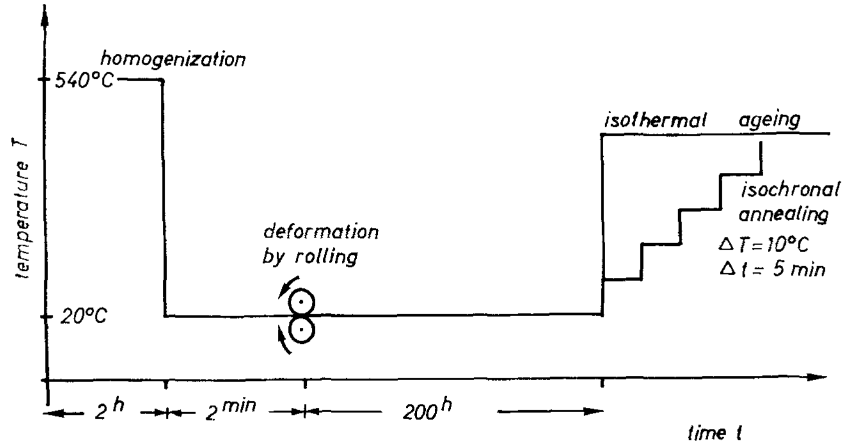
(ii) Formation and motion of high-angle boundaries of high mobility that act as a reaction front by sweeping out the dislocations ahead of their direction of motion.

The process (ii) is usually known as recrystallisation, or more specifically, *discontinuous recrystallisation*. Process (i) is termed recovery or *in situ* recrystallisation; the term *continuous recrystallisation* is better. Which of these processes takes place depends on the density and distribution of dislocations, as well as on segregation of solute atoms, and especially on the presence of particles connected with the dislocations.

The purpose of this work is to find the conditions under which the two different mechanisms take place in aluminium-copper alloys and whether there is a correlation between the annealing texture and the microscopic mechanism of recrystallisation. The experiments were done with alloys quenched from a temperature at which homogeneous solid solutions exist, deformed by rolling, and then heat-treated to obtain different precipitates, interacting with dislocations and grain-boundaries that have the tendency to anneal out. This work is based on former investigations on the effect of lattice defects on the formation of the various metastable and stable phases [2–5], and on the effect of precipitation on recrystallisation [6–8] in aluminium-copper alloys, and is related to the work on the formation of annealing textures in aluminium-iron alloys [9–13].

2. Materials and Procedures

Aluminium-copper alloys containing 1.95 and 5.00 wt % Cu (0.85 and 2.19 at. % Cu), prepared as previously described [3], were treated according to the following scheme:



According to earlier investigations [7] the heat-treatments were chosen so that either mechanism (i) or (ii) or both simultaneously determined the microstructure of the specimens. The amount of deformation and the heat-treatments for the specimens are recorded in table I.

TABLE I

Alloy	Deformation	Heat-treatment	Microstructure	Texture
Al–2 wt % Cu	90%	20° C	as rolled	rolling
Al–2 wt % Cu	90%	isochronal up to 310° C	mixed: cont. and discont.	mixed
Al–2 wt % Cu	90%	30 min 320° C	discont. recryst.	cube
Al–2 wt % Cu	90%	10^4 min 200° C	cont. recryst.	—
Al–2 wt % Cu	90%	10^4 min 280° C	discont. recryst.	—
Al–5 wt % Cu	50%	10^4 min 280° C	cont. recryst.	rolling
Al–5 wt % Cu	90%	10^4 min 280° C	cont. recryst.	rolling
Al–5 wt % Cu	90%	10^4 min 240° C	cont. recryst.	rolling

The changes in structure were investigated by transmission electron microscopy, selected area diffraction, and pole figure determination by X-rays. Half-quantitative {111} pole figures were determined by the back-reflection technique with a Siemens-texture-diffractometer and a scintillation counter, using CuKα radiation.

Electron microscopy was carried out on texture specimens using standard electrothinning procedures. A Siemens Elmiskop I operated at 100 kV was used for transmission electron microscopy and diffraction.

3. Experimental Results

3.1. Microstructures

The recrystallisation behaviour of aluminium-copper alloys deformed 90% by rolling is shown in fig. 1. Homogeneous solid solutions and alloys with relatively small supersaturation recrystallise discontinuously while highly supersaturated alloys recrystallise continuously. The boundary is not sharp. In its neighbourhood, both modes co-exist with volume portions that depend on annealing temperature and time. For isochronal heating, the transition boundary is shifted to higher temperatures because continuous recrystallisation starts first and is only replaced by the discontinuous process after particles that have formed at lower temperatures are partially dissolved during continued heating. With very low rates of heating, discontinuous recrystallisation can be avoided completely. The principal features of the two processes are discussed in terms of schematic drawings (figs. 2, 3) and a series of electron micrographs (fig. 4). The situation is somewhat complicated by the metastable θ' phase which forms at dislocations

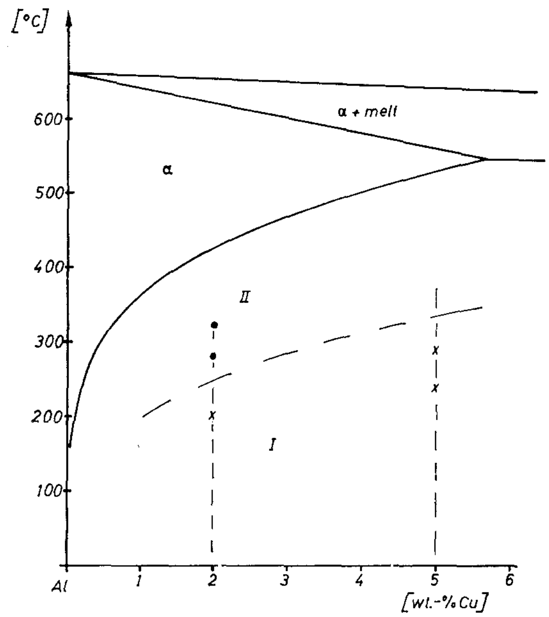


Figure 1 Phase diagram and recrystallisation behaviour of Al–Cu deformed 90% by rolling (I, continuous recrystallisation, II, discontinuous recrystallisation).

and sub-boundaries in addition to the stable θ phase which forms at grain-boundaries and at sub-boundaries below a certain dislocation spacing ($\sim$ 40 Å).

3.1.1. *Process (i) (figs. 2 and 4)*

After 90% plastic deformation by rolling, the sample contains a high density of dislocations ($N \sim 10^{12}/\mathrm{cm}^2$). In microband regions (i.e. zones of high misorientation between two crystal blocks produced by plastic deformation) the density is still higher. After heating above 100° C, θ phase nucleates at all original grain-boundaries and in the microband region the boundaries become pinned by a high density of equilibrium phase particles [14]. In the other areas θ' phase precipitates, and dislocations rearrange to form sub-boundaries (fig. 4a). Large θ' particles grow and finally transform into θ phase while small ones dissolve (fig. 4b). If a sub-boundary node is free of a particle it can move more easily and a sub-grain can anneal out by Y-node motion (fig. 2b). Another process involves a reduction

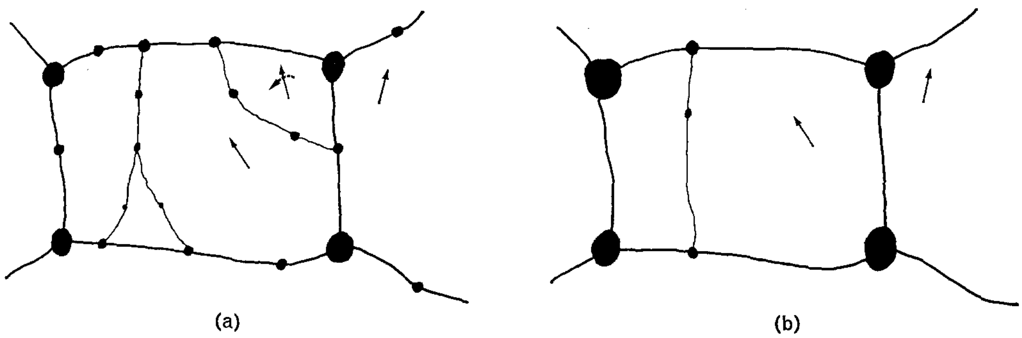
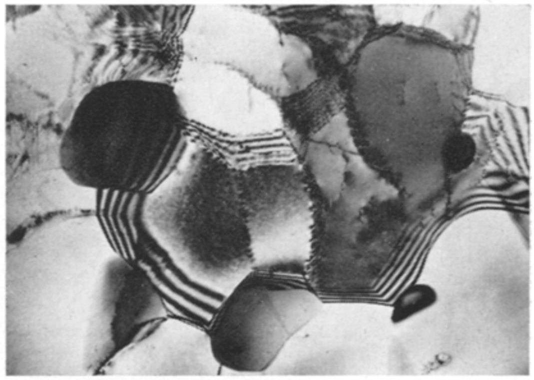


Figure 2 Schematic sketch of continuous recrystallisation. (a) Sub-boundaries are pinned by θ particles. (b) After dissolution of the smallest particles, a sub-grain can anneal out by Y-node motion (left) or by sub-grain rotation (right).

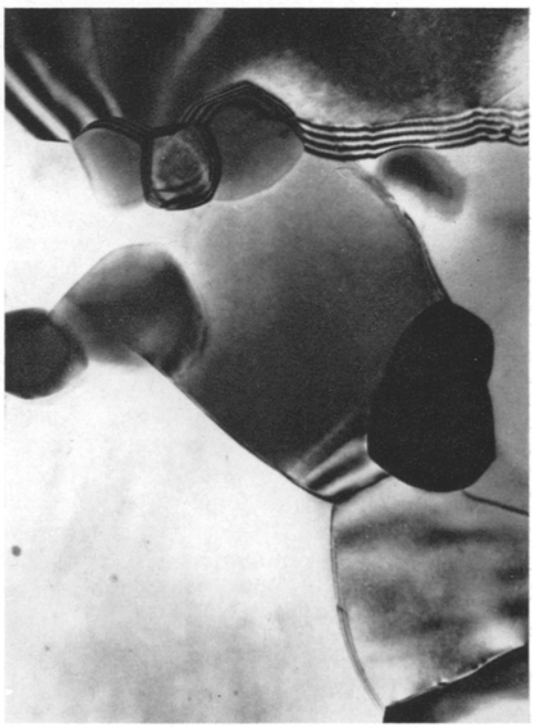
Figure 3 Schematic sketch of discontinuous recrystallisation. (a) The recrystallisation front sweeps out dislocations and sub-boundaries. (b) The recrystallisation front sweeps out dislocations and sub-boundaries and also transforms θ' to θ.

(a)

(b)



(c)



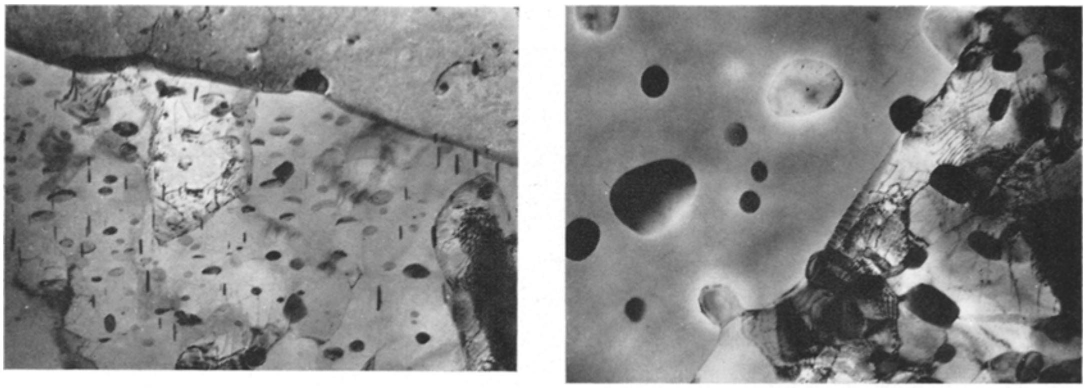
(d)

Figure 4 Continuous recrystallisation. (a) Dislocation networks between θ' particles: Al–3% Cu, 25% def, isochronal up to 370° C (× 80,000). (b) Al–5% Cu, 10% def, isochronal up to 390° C (× 21,400). (c) Al–5% Cu, 50% def, 10^4 min, 280° C (× 24,500). (d) Al–5% Cu, 90% def, 10^4 min 240° C (× 29,000).

of the dislocation density of a sub-boundary which implies rotation of a sub-grain [15, 16], (figs. 2b, 4c). Both reactions lead to a gradual increase in sub-grain size and to an increase in the misorientation between sub-grains. If the process takes place over a sufficient period of time a microstructure originates which, on a submicroscopic scale, is identical with one developed by discontinuous recrystallisation, i.e. dislocation-free crystallites of the equilibrium phase and grain-boundaries (fig. 4d).

3.1.2. Process (ii) (figs. 3 and 5)

At low supersaturation, fewer particles precipitate at original grain-boundaries and in microband regions, so that some of these can start to move as recrystallisation fronts. The driving force for this reaction comes from the dislocations and from the transformation of metastable θ' into θ (fig. 5a). This is valid for a moderate degree of deformation. At high deformations, nucleation sites for θ exist in such large numbers that practically no θ' phase is present when the recrystallisation front arrives (fig. 5b). Recrystallisation then takes place in a matrix in which the solute concentration has been decreased for example from 5 wt % to 0.2 wt % Cu if aged at 250° C. The dispersed particles must however



(a) (b)

Figure 5 Discontinuous recrystallisation. (a) Al–2% Cu, 90% def, isochronal up to 310° C (× 15,300). (b) Al–5% Cu, 50% def, isochronal up to 400° C (× 15,300).

fulfil the condition that their retarding force is smaller than the driving force due to the dislocations [14].

3.2. Preferred Orientations

Samples which had been recrystallised by either of these two modes of recrystallisation, or contained a mixture of both types, were investigated by selected area electron diffraction and by X-rays to determine the distribution of orientations. With the first method, areas of a specimen were selected in which discontinuous or continuous recrystallisation had taken place. A high frequency of {100} orientations was found if the discontinuous process had taken place, while after continuous recrystallisation most of the regions were found to have {211} or {110} planes parallel to the sheet surface.

The results of the X-ray pole figure determinations are listed in table I. The three important types of annealing textures are shown in fig. 6b, c and d, in addition to the original rolling texture (fig. 6a). The results indicate that the rolling texture, which consists mainly of (112) [111] and (011) [211] components, is completely preserved if the continuous process takes place, while the cube texture (001) [100] is formed by the discontinuous process. Mixed microstructures lead to textures with mixed components. These results seem to be applicable to the interpretation of textures in a large number of heterogeneous alloys.

4. Discussion

The relationship between the microscopic mechanism of recrystallisation and texture can be interpreted as follows. In the case of continuous recrystallisation, no reorientation through large angles can take place. Reorientation that takes place on a microscopic scale does not lead to a change in texture as determined by the macroscopic X-ray method. The cube texture is the annealing texture of rather high purity aluminium, as it is for other high purity, high stacking fault energy fcc metals. Homogeneous solid solutions of these metals show increasing randomness of the annealing texture with increasing solute content. The annealing texture of the homogeneous Al–Cu solid solutions used in this investigation should be random. The reason for the occurrence of the cube texture must be the precipitation of the copper in front of the reaction. An equilibrium with the stable θ phase leads to the lowest possible copper concentration at any given annealing temperature. There is minimum segregation of copper and therefore a strong anisotropy of grain-boundary mobility, just as in a homogeneous material of corresponding purity. This behaviour can only be expected if the particles are thermodynamically stable towards the grain-boundary. If this is not the case particles dissolve in a passing grain-boundary (as observed in Ni + Ni_3Al alloys), and a high degree of segregation and no preferred orientation is to be expected for the annealing texture [1].

Acknowledgement

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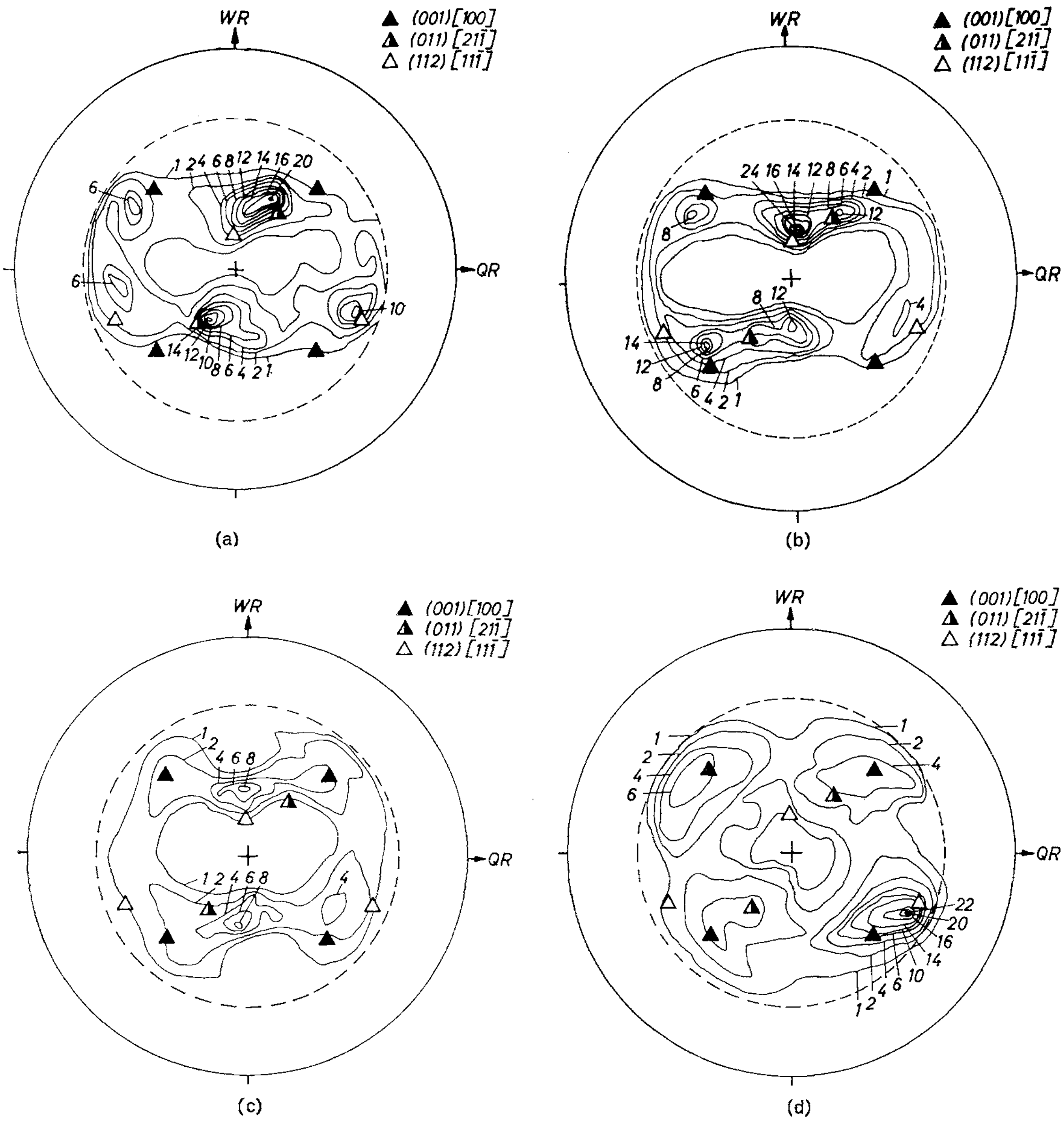

Figure 6 {111} Pole figures. (a) Al–2% Cu, 90% def, 20° C, rolling texture. (b) Al–5% Cu, 90% def, 10^4 min, 280° C, annealing texture. (c) Al–2% Cu, 90% def, isochronal up to 310° C, annealing texture. (d) Al–2% Cu, 90% def, 30 min, 320° C, annealing texture.

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