JOURNAL OF MATERIALS SCIENCE 6 (1971) 1350–1354

On the Mechanism of Feathery Crystallisation of Aluminium

HASSE FREDRIKSSON, MATS HILLERT
Royal Institute of Technology, Stockholm 70, Sweden

The growth of twinned crystals of aluminium on solidification is discussed with reference to the structure of the growing tip. It is shown that one must differentiate between two types of ⟨112⟩ growth directions in the twin boundary. Normally, all the tips grow in the same type of ⟨112⟩ growth direction and this particular type may give the tip a favourable shape, possibly by producing a sharp edge. If a tip is growing in a less favourable ⟨112⟩ direction, the favourable type can be established by the twins changing sides. This is observed to occur regularly in the continuous casting method according to Hunter. The re-entrant edge mechanism is not supported by the present results.

1. Introduction

The solidification of semicontinuously cast ingots of aluminium alloys frequently results in a large number of parallel, thin lamellae, usually called feathery grains. Each lamella consists of two crystals in twin orientation, the twin boundary (111) being located along the central plane of the lamella. It is self-evident that each pair of twins has formed by growth side by side in some crystallographic direction in the common twin boundary and examination by means of an etch-pit method [1] and ordinary X-ray methods [2] has shown that growth occurs preferentially in a ⟨112⟩ direction.

For an isolated crystal, there are six equivalent ⟨112⟩ directions in each (111) plane. For a pair of twins, however, these six directions fall into two different groups with, for instance, $[11\bar{2}]$, $[1\bar{2}1]$ and $[\bar{2}11]$ belonging to one group and $[2\bar{1}\bar{1}]$ $[\bar{1}\bar{1}2]$ and $[\bar{1}2\bar{1}]$ belonging to the other. The difference can be illustrated by the atomic arrangement in a plane perpendicular to the (111) plane (fig. 1). This difference has previously been neglected in the discussions of the formation of the feathery grains and will be the main theme of the present communication.

2. Identification of Preferred Growth Direction

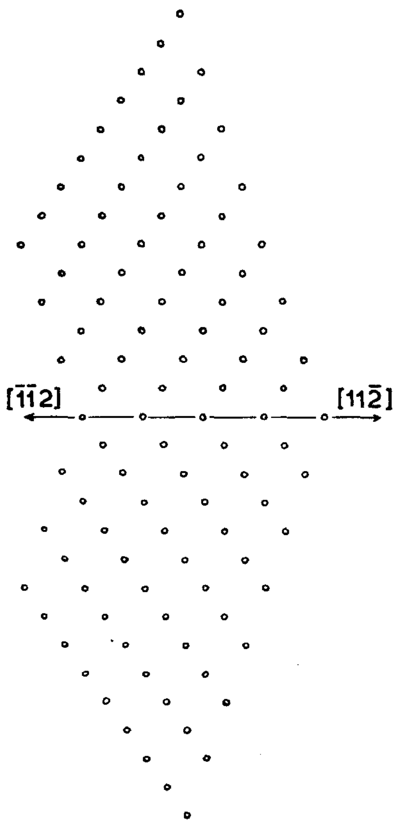

Figure 1 The atomic arrangement of a pair of twinned crystals in a plane perpendicular to the twin boundary. The diagram is bounded by sections of {111} type planes.

Microscopic examination by means of the etch-pit method [1] or polarised light after anodic oxidation [3], has revealed that a large area may be covered by alternating plates of two crystals in twin orientation. The boundaries between the plates are alternatingly straight and wavy, and decanting experiments [3] have shown that the straight boundaries form at protruding tips or edges, whereas the wavy boundaries form as the neighbouring lamellae meet by side-ways growth (fig. 2). Each pair of plates with a straight boundary in the middle can be regarded as one lamella and resembles a feather in the plane of section.

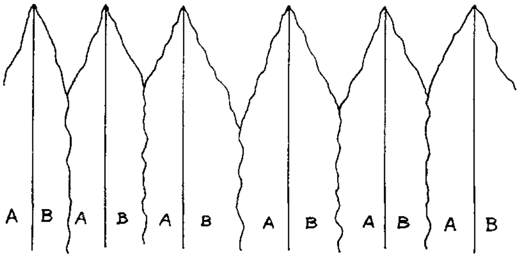

Figure 2 Illustration of the growth of twins leading to two different types of twin boundaries.

By examining a large number of specimens we have observed that the same arrangement is always found. The crystal called A in fig. 2 is always on the left-hand side of the straight boundary and the crystal B is always on the right-hand side. This observation indicates that there is an important crystallographic difference between the straight and the wavy boundaries. This difference may be explained with reference to fig. 1. If one of the boundaries has the structure illustrated at the $[1\,1\,\bar{2}]$ vector, the next boundary must have the structure illustrated at the $[\bar{1}\,\bar{1}\,2]$ vector and so forth. This fact is illustrated in fig. 3.

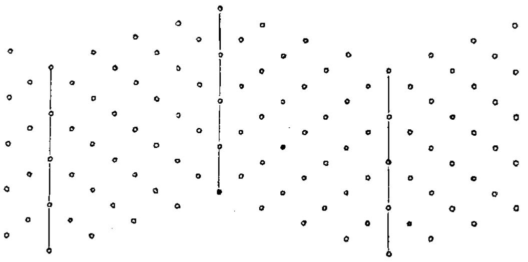

Figure 3 The atomic arrangement at the two types of twin boundaries.

In order to determine which kind of boundary is straight and which one is wavy, we have made an examination with polarised light after anodic oxidation of a specimen sectioned in the same way as in figs. 1 to 3. A stress-free objective was selected and the polars were crossed, the polariser in the horizontal position and the analyser adjusted to the vertical position. The specimen was then rotated on the microscope stage. The result is presented in figs. 4 to 6, where the direction of solidification is identified with an arrow. This sequence repeated itself four times during each revolution as expected with crossed polars and a stress-free objective.

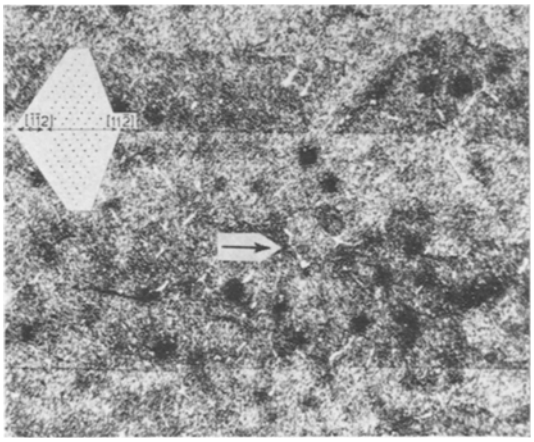

Figure 4 The two twins show the same brightness under crossed polars when the twin boundary is parallel to the plane of polarisation. The arrow shows the direction of solidification. The plane of polarisation is indicated by the sign ↔ and the analyser is in the perpendicular position (× 200).

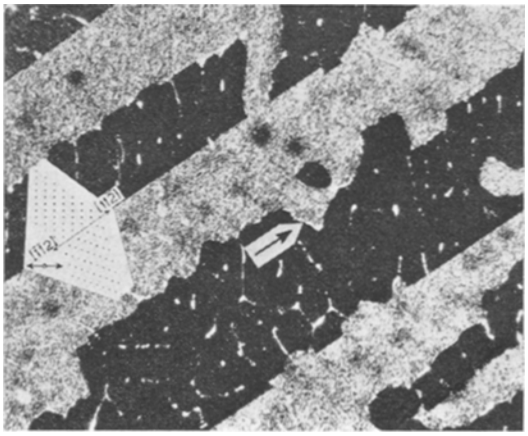

Figure 5 One twin shows maximum darkness at 35° inclination. The arrow shows the direction of solidification. The plane of polarisation is indicated by the sign ↔ and the analyser is in the perpendicular position (× 200).

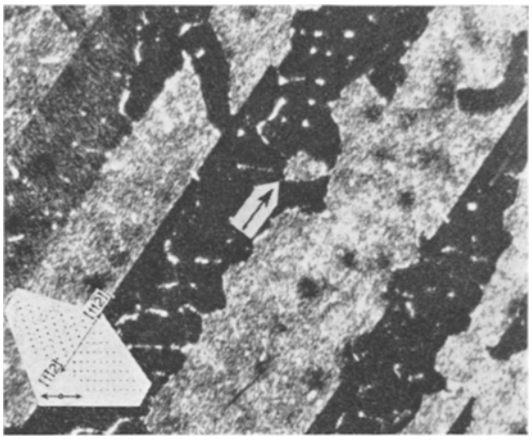

Figure 6 One twin shows maximum darkness at 55° inclination. The arrow shows the direction of solidification. The plane of polarisation is indicated by the sign ↔ and the analyser is in the perpendicular position (× 200).

As shown by the sequence of pictures, one of the crystals appeared dark when the twin boundary was at an angle of about 35° to the horizontal line and the other crystal was dark at an angle of about 55°. This is expected from the symmetry of the atomic arrangement as shown to the left of each picture. In the plane of section, the atoms are arranged on lattice points at the intersections of two sets of perpendicular lines. Such a surface will be optically neutral when the crossed polars are parallel to these lines and the crystal will then look dark. By comparing the pictures with the atomic arrangements we may thus conclude that the straight twin boundaries are formed by favoured growth in a ⟨112⟩ direction such as [11$\bar{2}$] in fig. 1 whereas the wavy twin boundaries have the properties demonstrated by [$\bar{1}\bar{1}$2] in fig. 1.

3. Discussion of the Rate-controlling Mechanism

The ⟨100⟩ direction is usually the preferred growth direction for dendritic growth of individual aluminium crystals. There are two lines of thought regarding the explanation of this fact. According to one, the process of atomic attachment is kinetically favoured on the {100} surface. According to the other, the geometry of the tip of a dendrite-arm would be most favourable for the diffusion of heat or solute atoms when it grows in the ⟨100⟩ direction because of an anisotropy in surface energy. It may be expected that the close-packed {111} surfaces have the lowest surface energy and the fact that a [100] direction is surrounded by four {111} planes should then result in a tip in the [100] direction if the geometry is controlled by surface energy. This is most evident in fig. 7a where a very strong anisotropy has been assumed but it is also evident in fig. 7b where a less pronounced anisotropy was assumed.

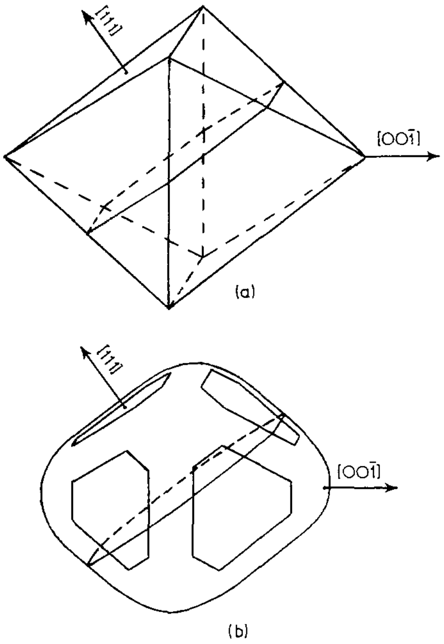

Figure 7 Equilibrium shape of isolated crystal (a) with high degree of anisotropy and (b) with a lower degree of anisotropy.

Fig. 8 illustrates the corresponding situation for a twin arrangement. It is obtained by rotating the lower part of the crystal in fig. 7 around the [111] axis. In order to facilitate comparison, the [111] vector and plane are indicated already in fig. 7. It should be noticed that fig. 8 is realistic only if the surface energy of the twin boundary is negligibly small and this is not quite true. With this reservation, we can draw the following conclusions from fig. 8. If a pair of twins forms a tip in a [112] growth direction in the twin boundary, the tip may obtain a sharp edge if it is growing in a direction equivalent to [11$\bar{2}$] in figs. 1 and 8 and it may obtain a re-entrant edge if it is growing in a direction equivalent to [$\bar{1}\bar{1}$2].

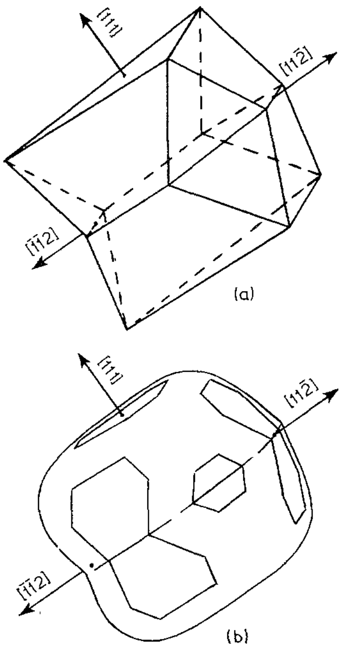

Figure 8 Equilibrium shape of twinned crystal assuming negligible twin boundary energy (a) with high degree of anisotropy and (b) with a lower degree of anisotropy.

Re-entrant edges may be of considerable importance for the growth of crystals when the atomic attachment process is rate controlling and it is interesting that Chalmers [4] suggested the re-entrant edge mechanism to be operative in the growth of feathery grains of aluminium. However, our observation that the favoured growth direction is of the type represented by [11$\bar{2}$] in figs. 1 and 8 seems to rule out the re-entrant edge mechanism. Furthermore, the fact that fig. 8 predicts the formation of a sharp edge for the favoured growth direction seems to support the importance of the geometric shape of the tip. It thus seems reasonable that the competitition between various growth directions for dendritic and feathery growth of aluminium under ordinary casting conditions is governed by the principle, "the sharper the tip, the higher is the growth rate".

4. Observations on Feathery Growth in Continuous Casting according to Hunter

A particularly interesting case of feathery growth of aluminium is sometimes found in the continuous casting method according to Hunter [5]. The material is cast directly in a rolling mill and solidifies as a thin sheet between two rotating rolls. When the material solidifies in the feathery fashion, it has been found [6] that the growth direction in the outer layer of the sheet is not ⟨112⟩ but a direction rather close to ⟨110⟩ which is also in the twin boundary. Furthermore, the outer layer recrystallises before it is cooled and the recrystallisation takes place by one of the twins consuming the other. Strangely enough, the roles seem to change every centimetre. First one twin consumes the other one for about 1 cm, then the other twin consumes the first one for about 1 cm and so on. In view of the results presented in the previous paragraphs, it is tempting to suggest the following explanation.

The material which has just solidified between the rolls is immediately deformed somewhat by them. This may give rise to a continuous rotation of the pair of twins growing in the feathery fashion. The growth can continue even if the favoured ⟨112⟩ direction is thus moved away from the direction of solidification. The growth direction rotates towards a ⟨110⟩ direction, it passes this direction and starts to rotate towards the next ⟨112⟩ direction which is of the wrong type. However, by changing sides at the growing tips, the two twins can establish a twin boundary rotated 180° from the previous one and the new one will thus be of the favourable type. It seems that the direction of rotation is now reversed and the growth direction again passes through the ⟨110⟩ direction.

The hypothesised changing of sides can in fact be observed in the interior of the sheet where the material apparently has less time to recrystallise. Fig. 9 gives an example.

The fact that one twin consumes the other one may simply be due to an initial difference in volume fraction, the more abundant twin being

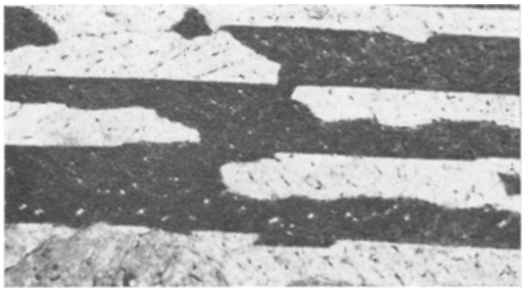

Figure 9 Micrograph showing how the twins change sides (× 400).

the winner. An initial difference in volume can be expected if the direction of solidification deviates somewhat from the twin boundary. The twin which is situated on the favourable side of the twin boundary will grow thicker than the other one. After the twins have changed sides, the other one will grow thicker and later on will come out as the winner.

Acknowledgement

Thanks are due to the staff of the research laboratory at Svenska Metallverken in Finspång and in particular to Mr Rolf Sundberg for stimulating discussions and valuable information especially on the "Hunter material", and for supplying us with material for this investigation.

References

1. J. HÉRENGUEL, *J. Metals* **4** (1952) 385.
2. J. KARP and J. SZYMANSKI, *Rudy i Metale Niezelne* **11** (1966) 245.
3. L. R. MORRIS, J. R. CARRUTHERS, A. PLUMTREE, and W. C. WINEGARD, *Trans. AIME* **236** (1966) 1286.
4. B. CHALMERS, "Principles of Solidification" (J. Wiley and Sons, New York, 1964).
5. D. ALTENPOHL, *Metall* **20** (1966) 234.
6. R. SUNDBERG, Internal Report (Svenska Metallverken, Finspång, Sweden, 1967).

Received 25 May and accepted 28 July 1971