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Transition bands and recrystallization in metals

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[Plate 30]

Transition bands are regions of high lattice distortion which result from inhomogeneous deformation in metals. During annealing of cold-worked metals, nucleation of recrystallization is frequently observed to occur within these bands.

An analysis has been made of some of the conditions under which transition bands may develop during deformation, and of the factors which influence their microstructure. A model is proposed for the mechanism by which nucleation of recrystallization occurs in transition bands during annealing. The dependence of these processes on crystallographic orientation is then discussed and related to the formation of recrystallization textures in b.c.c. metals.

1. INTRODUCTION

When a cold-worked metal is annealed at temperatures greater than about one third of its melting point on the absolute scale, the process of recrystallization occurs and gives rise to many important changes in properties. The very great reduction in dislocation density during recrystallization causes considerable loss in strength, but there are associated increases in ductility and formability. These latter properties are widely exploited in many applications of annealed steel sheet and other metal products. During hot working of steel, recrystallization concurrent with deformation has the beneficial effect of removing work hardening and thereby reducing the work load, and maintaining malleability.

Initiation of recrystallized grains is known to occur in two ways. The first, which was studied by Beck & Sperry (1950) and later by Bailey & Hirsch (1962) involves the migration of an existing high angle grain boundary. This mechanism has been examined in some detail by the latter workers and can be explained in reasonably quantitative terms. Alternatively, recrystallization may start in the deformed structure within grains by growth of a small strain-free region. This concept was advanced by Beck (1949) and Cahn (1950) and is widely accepted, although the precise mechanism of growth remains in some dispute. It is generally agreed that this process occurs most readily in regions of high local deformation and sharp lattice curvature. Such regions have been termed 'transition bands' by Walter & Koch (1963) and have been shown to act as favoured sites for recrystallization by

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these workers and also by Hu (1963). However, there seems to have been no systematic examination of the conditions under which regions of sharp lattice curvatures may be created during deformation. The present study was carried out to determine some of these conditions by application of macroscopic plasticity theory. In addition, the mechanism by which recrystallized grains can develop within transition bands is analysed in some detail.

2. Origins of transition bands

Transition bands arise when the deformation of a crystal is inhomogeneous such that it breaks down into separate fragments which are mutually misoriented. These fragments, which were termed deformation bands by Barrett & Levenson (1940), may be observed in optical metallography by a surface orientation dependence of etching. Additionally, the transition bands which accommodate the misorientation between adjacent deformation bands may themselves be visible. Boundaries of this type in deformed single crystals of silicon iron have been examined using the electron microscope by Hu (1963) and Walter & Koch (1963). Figure 1, plate 30, is an electron micrograph showing a transition band in a polycrystalline iron specimen which had been cold rolled 50 %. It is seen to consist of many elongated subgrains with their long axes almost parallel. The misorientation across the transition band can be determined from the accompanying diffraction patterns which, in this case, show a deviation of 30° over a distance of only 3 μm.

Deformation bands may arise in four ways, as follows:

(i) if, as a result of local differences in stress state, different parts of the crystal undergo different strains there will in general be some relative rotation;

(ii) if the strain throughout a crystal is specified in all its components, and is homogeneous it is still possible that different combinations of slip systems may separately yield the imposed strain in different parts of the crystal, giving different slip rotations and causing relative rotation;

(iii) a macroscopically imposed shape change may be achieved for a lower energy expenditure if the total strain is subdivided such that different regions undergo different strains, and again rotate relative to each other;

(iv) a crystal may be oriented in such a way that it does not rotate when subjected to a specified strain, but that small displacements in opposite senses from the initial orientation would cause the two orientations to continue to diverge.

Deformation bands arising as in (i) will invariably occur in polycrystals and are likely to form at low strains and develop progressively with strain. At increasing strains the rate of formation of bands from this cause will diminish as deformation becomes more homogeneous (Budiansky & Wu 1962). The likelihood of banding as specified in (ii) depends on the number of independent slip modes. In f.c.c. metals where slip occurs on $\{111\}$ planes in $\langle 110\rangle$ directions there are twelve slip systems. With the assumption of homogeneous deformation, each crystal orientation offers the possibility of choice of several different combinations of five independent slip

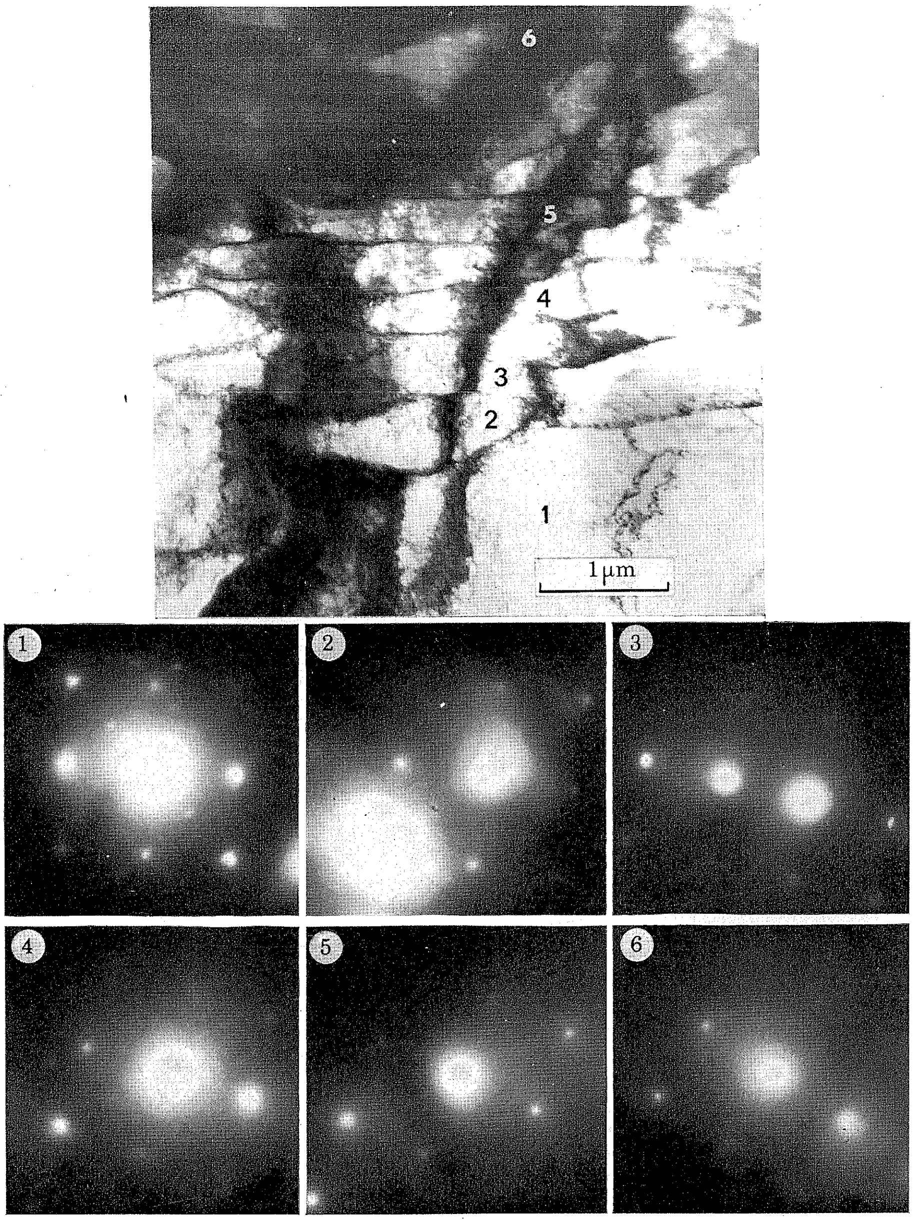


Figure 1. Electron micrograph showing a transition band in iron rolled 50 %. The misorientation across the band is seen from the diffraction patterns.

systems from the six or eight systems, which are equally stressed under one of the Bishop & Hill (1951) stress states. The same would be true for b.c.c. metals deforming on $\{110\}\langle 111\rangle$ systems. For b.c.c. metals deforming in pencil glide on $(hkl)\langle 111\rangle$ systems there is, except in a few symmetrical cases, only one possible combination of slip systems to yield the imposed strain.

Case (iii) is in the same category as case (i) but case (iv) is quite different and, from the viewpoint of analysis, has the advantage over other forms of banding of being readily predictable.

In the present work, an attempt will be made, initially, to explore the conditions under which a transition band is likely to provide nuclei for recrystallization. The model of transition band nucleation will be assessed by examining the way in which orientations that are expected to form deformation bands as in case (iv) are developed or arise during deformation, and by comparing the proposed behaviour with information derived from a study of recrystallization textures.

3. CONDITIONS FOR NUCLEATION WITHIN TRANSITION BANDS

Figure 2 is a schematic diagram of a transition band in two dimensions. The axis of rotation in the band lies in the plane of the paper in the direction indicated, it being assumed here that the axis of rotation is fixed. This assumption is not limiting but is chosen as being consistent with the simplest possible model which represents accurately the important parameters.

There are two types of small angle dislocation boundaries, which may be identified in figure 2. Adapting the terminology used by Ashby (1970), we can say that the boundaries are composed of geometrically necessary dislocations and statistically stored dislocations.

The geometrically necessary dislocations are those which accomplish the rotation across the transition band while the statistically stored dislocations give random misorientations. Two types of dislocation boundaries are apparent in figure 1, those separating the lath subgrains, which are designated as transition boundaries and are composed primarily of geometrically necessary dislocations, and the more widely spaced boundaries approximately perpendicular to the transition boundaries, which are composed primarily of statistically stored dislocations and will be referred to as random boundaries. In figure 2 the transition boundaries are represented as being of tilt character but this is not of importance to the present discussion. The mean spacing and specific energy of the transition boundaries are d_t and σ_t and for the random boundaries the corresponding parameters are d_r and σ_r.

A potential recrystallization nucleus is invariably a subgrain which, at the critical stage in its development is distinguished from its neighbours by a size advantage. The necessary size advantage depends on geometry and boundary energy considerations (Bailey & Hirsch 1962). Both the length and thickness of the laths are important dimensions governing nucleation within transition bands, not solely the lath thickness as assumed by Hu (1969). In order to define the important parameters more closely consider the simplified configuration shown in figure 2. The large subgrain of length D_r exists in a matrix of subgrains of average length d_r and width d_t. In the deformed state, the sub-boundaries meeting at junctions such as the points a–k in figure 2, do not lie at the angles relative to each other which would be dictated by the sub-boundary tensions. On annealing relaxation will occur at these junctions as the sub-boundaries seek to adjust their position to equalize the surface tension forces. The subgrain of length D_r is considered to be a potential nucleus because it has a free segment, bd, which can move towards the boundary ce as a result of the tension exerted by the two transverse boundaries. The structure

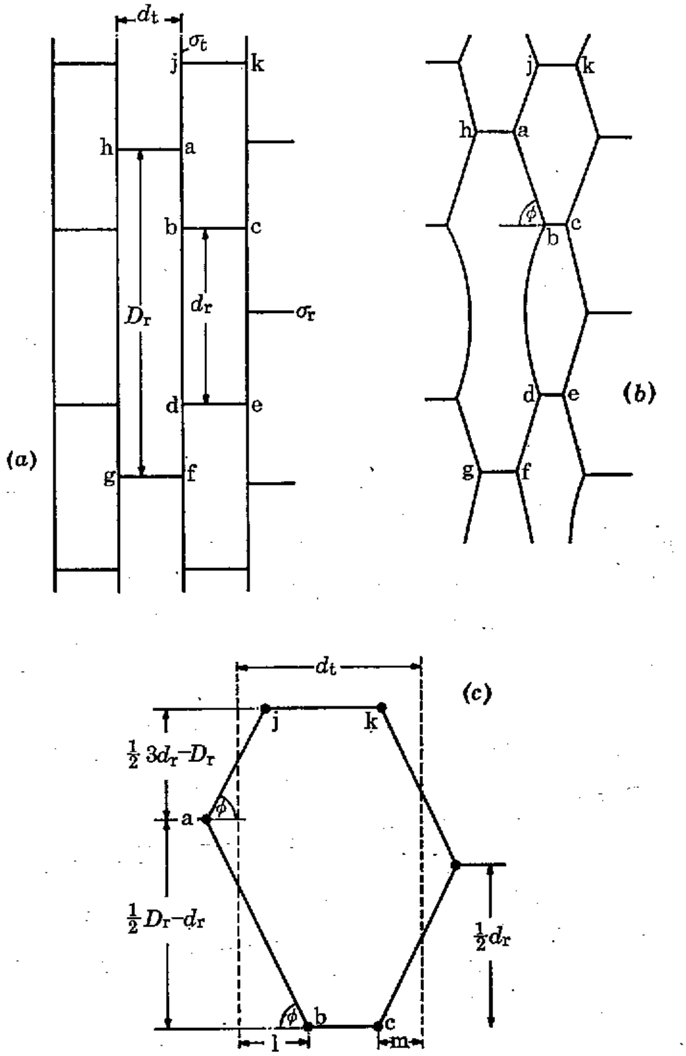


FIGURE 2. Schematic representation of a transition band, (*a*) as formed, (*b*) relaxed on annealing, (*c*) detail of the potential nucleus.

shown in figure 2*a* will relax initially to the configuration shown in figure 2*b*. This figure has been drawn to indicate the position at which the boundary angles first reach the equilibrium of boundary tension, the angle ϕ being given by

$$\cos\phi = \sigma_r/2\sigma_t,$$

where σ_r is the specific boundary energy of the random boundaries, such as bc and de and σ_t is the energy of the transition boundaries, such as ab and df. The configuration of figure 2*b* is not, of course, in equilibrium because the boundary segment bd is curved. However, a simple indication of the underlying physics is given by considering that nucleation will have occurred if the points b and c meet each other at this stage, before the boundary tension in segment ab restricts its movement. The relaxed configuration is shown in greater detail in figure 2*c*, where the relevant boundary angles and lengths are indicated. The condition that points b and c should meet is given by

$$l+m \geqslant d_t,$$

which becomes

$$D_r \geqslant \frac{4}{3}\left(d_r + d_t\left[\frac{4\sigma_t^2}{\sigma_r^2} - 1\right]^{\frac{1}{2}}\right). \tag{1}$$

Although this model is very much simplified compared with the true situation it undoubtedly gives a roughly correct indication of the importance of the different parameters. It also allows a general indication of the nature of the nucleation process, showing that the growth of an individual subgrain is aided by, and, in some cases, competing with the overall tendency for the structure to spheroidize. It may be noted from figure 2 that if the points a and h come together the subgrain of length D_r will shorten, and all surrounding subgrains will broaden by the removal of alternate transition boundaries. In this process of spheroidization the enlarged subgrain will retain a size advantage and will continue to grow, but it will not enjoy so large an advantage as it would if spheroidization did not occur. If the points a and h do not meet, the subgrain of initial length D_r will broaden initially to the width $2d_t$ and subsequent sideways growth will be accompanied thereafter by longitudinal growth under a driving force derived from the transition boundaries. Thus the enlarged subgrain will have a high probability of continuing to satisfy the condition expressed in equation (1) by successive growth stages through the curved lattice as the magnitude of its boundary energy increases.

The condition that spheroidization should not occur around the subgrain of length D_r is given by

$$D_r \geqslant 3d_r - 2d_t\left(\frac{4\sigma_t^2}{\sigma_r^2} - 1\right)^{\frac{1}{2}}. \tag{2}$$

The two relations (1) and (2) are shown in figure 3, where D_r is plotted against θ_t, the angle across individual transition boundaries. The curves of negative slope represent the boundary condition for spheroidization, those of positive slope give the minimum value of D_r for nucleation. The full lines, all having a cusp at the value $D_r = 2.0$ represent the condition that nucleation should occur without spheroidization. There are clearly contrary requirements for inhibiting spheroidization and promoting nucleation. Nucleation is favoured by narrow laths with low misorientation across successive boundaries; spheroidization is inhibited by widely spaced boundaries of high misorientation.

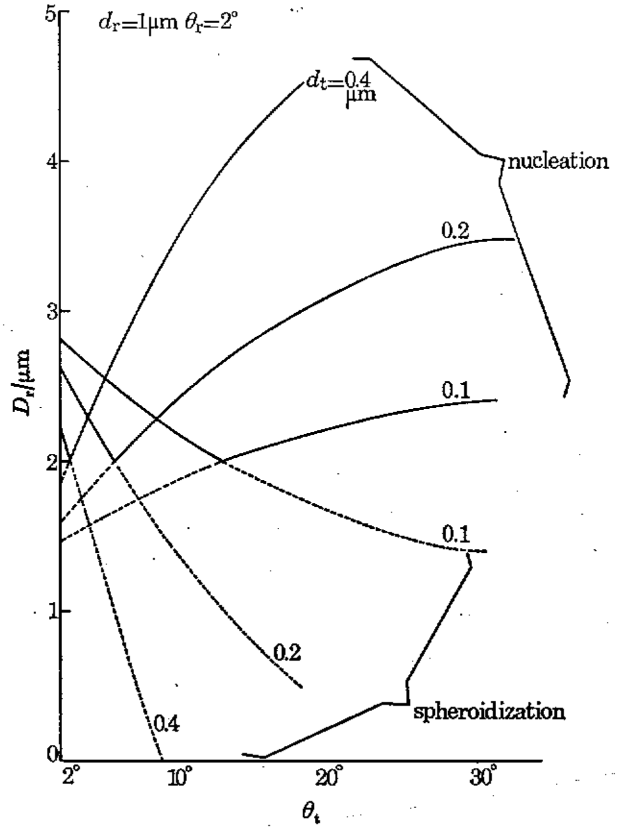


FIGURE 3. Conditions for nucleation and spheroidization.

However, it is only necessary to consider the inhibition of spheroidization if the most highly favoured case of nucleation is to be considered. In reality the nucleation condition of equation (1) will be the more important of the two. Nucleation will not occur at essentially zero time, however, for subgrains which satisfy equation (1) but not equation (2). An incubation period will be found corresponding to the duration of the subgrain growth and spheroidization during which the enlarged subgrain reaches sufficient size to break away.

It is clear from the variation of D_r with θ_t shown in figure 3 that narrow subgrains of small successive misorientations also provide the most favourable conditions for continued growth following nucleation. As the nucleating subgrain widens, its boundary misorientation increases, and figure 3 indicates the rate of change in D_r necessary to sustain growth in this situation. The possibility that d_r and θ_r may vary with orientation will be considered later, but it is obvious that if d_r decreases and θ_r increases away from the nucleus site, the necessary rate of increase of D_r for sideways growth will be reduced.

The curvature in a given transition band is defined by d_t and θ_t, but, although it is possible to calculate the curvature of the lattice as a function of strain (from a known or assumed initial curvature) it is not possible to separate the variables d_t and θ_t theoretically. There is experimental evidence, however (Hu 1969), which suggest that d_t approaches a limiting value with increasing strain, so that the misorientation across individual sub-boundaries must increase correspondingly as the curvature becomes sharper. In this case the size advantage required by a potential nucleus increases so that a given transition band may cease to be a viable site for nucleation of recrystallization after large strains. Again, at low strains a transition band will not be sufficiently developed (large d_t) to satisfy equation (1) and will have too low a total misorientation to be effectively distinguished from a region of small, random misorientation.

It must not be forgotten that the especial virtues of a transition band for nucleation of recrystallization are that (*a*) individual sub-boundaries are of low angle so that the critical nucleus size is small, and (*b*) a small growth range allows the development of a high angle boundary so that the low mobility of small angle boundaries is not an abiding limitation.

Returning to equation (1), it is of value to examine further the parameters involved, and to direct attention to the values of d_r and σ_r which depend primarily on the statistical storage of dislocations. It may especially be noted that the statistical storage of dislocations is a function of orientation (Dillamore, Smith & Watson 1968) so that it is possible that there will be a gradient of both d_r and σ_r across a transition band. The next section will introduce some experimental results to illustrate this point, and indicate how these parameters vary with orientation.

4. THE ORIENTATION DEPENDENCE OF SUBSTRUCTURE

The results presented in this section are taken from earlier work (Dillamore, Smith & Watson 1968) and supplemented by some new results, all results being obtained from rolled pure iron. The technique by which these results were obtained is described in the earlier publication. Here it is intended to consider how subgrain size and sub-boundary misorientation depend on orientation parameters; the results to be considered are presented in figure 4. The orientation parameter in this case is the angle between the sheet plane normal and [001] for orientations having $\langle 110 \rangle$ parallel to the rolling direction. It must be remembered in considering the experimental results that they were obtained from polycrystalline material and orientations are determined from diffraction patterns. No information is available regarding the orientation history of a particular volume of material, although this may be expected to influence the development of substructure. It is not possible, therefore, to do more than indicate trends for the effect of orientation on substructure.

The first conclusion indicated by the results of figure 4 is the obvious one that the cell dimension decreases and the cell boundary misorientation increases with increasing strain. More important, however, is the fact that these same trends are observed as a function of orientation for a fixed rolling strain as the orientation parameter increases from 0° to 90°.

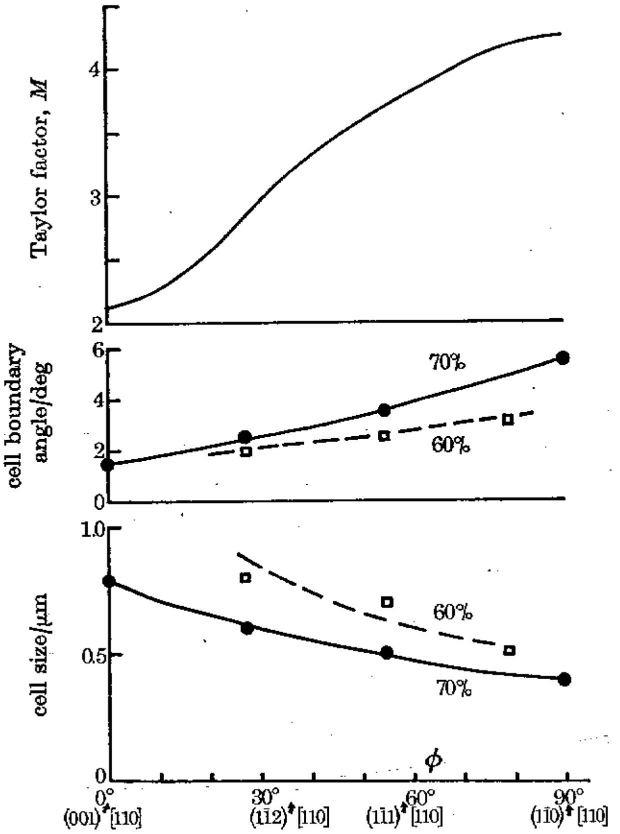


FIGURE 4. The variation of substructure in iron with orientation and degree of deformation.

The dependence of substructure on orientation clearly relates to the amount of dislocation movement, and possibly to the types of dislocation interaction. Little can be said in relation to the latter possibility except that different combinations of dislocations will be found in different orientations. In the cases considered here of b.c.c. crystals all rolled with a ⟨110⟩ rolling direction, the dislocations necessary to achieve the imposed strain will depend on the orientation of the rolling plane. Two, three, or four families of dislocations with different Burgers vectors may be involved depending on the crystal orientation. Since the orientations in the range considered here are very nearly stable during rolling deformation, any dislocations which remain may be categorized as statistically stored. The substructure formed is therefore of similar nature to the random boundaries in a transition band.

After heavy deformation, dislocations are distributed mainly in cell boundaries and so the dislocation density, f, relates approximately to the cell size and misorientation, through

$$f = \theta / b d_r,$$

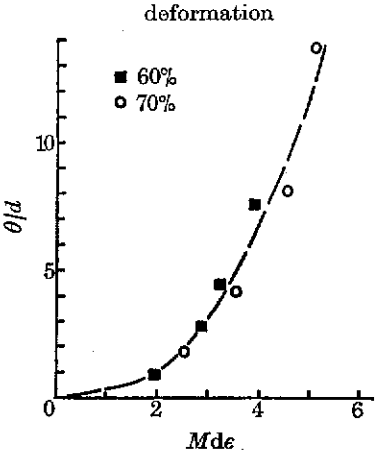


Figure 5. Data from figure 4 re-plotted as θ/d (a measure of dislocation density) against $M\mathrm{d}\epsilon$, the total shear strain.

where θ is the sub-boundary angle, b is the magnitude of the Burgers vector of the dislocations and d_r is the cell dimension. The dislocation density increases with total shear strain, and in producing a specified shape change, the total shear strain is a function of orientation described by the variation of the Taylor factor, M, defined as

$$M = \Sigma\, \mathrm{d}\gamma / \mathrm{d}\epsilon$$

where $\Sigma\, \mathrm{d}\gamma$ is the sum of the shears on the crystallographically defined slip systems, and $\mathrm{d}\epsilon$ is the largest principal strain. Hence the parameters θ and d_r are also functions of the Taylor factor M.

A crude estimate of the rate of dislocation storage as a function of strain may be obtained by determining f as a function of $\Sigma\, \mathrm{d}\gamma$. This is done in figure 5 where the results of figure 4 are re-plotted as θ/d_r (a measure of the dislocation density) against $M\,\mathrm{d}\epsilon$. Note that no account is taken of the change in orientation, and therefore in M, with strain.

The results presented here have established that the rate of dislocation storage is a function of the M value and that the cell boundary misorientation, θ, is larger and the cell boundary spacing, d_r, is smaller the larger the value of M. A cell boundary has an energy σ given approximately by the formula of Read & Shockley (1950)

$$\frac{\sigma}{\sigma_0} = \frac{\theta}{\theta_0}\left(1 - \ln \frac{\theta}{\theta_0}\right),$$

where θ_0 is the misorientation corresponding to the maximum boundary energy σ_0.

Associating the present discussion with the structure of transition bands, as the orientation changes across a transition band so also will M change. The random boundary spacing d_r and energy σ_r will be dependent to some extent on the local value of M and a structure gradient across a transition band will thus arise. The transition boundary spacing and misorientation will be determined mainly by other processes considered in the next section.

5. Transition band development

In this section we shall discuss the development of transition bands by rotations due to slip. For the purpose of calculation, the shear is considered to occur on $\{hkl\}$ $\langle 111\rangle$ slip systems, corresponding to pencil glide in b.c.c. structures. This choice of slip mechanism is appropriate to α-iron and other b.c.c. metals, and has the added virtue of allowing the deformation parameters to be unambiguously determined. That is to say that in determining M for an imposed shape change, the orientation of the $\{hkl\}$ slip plane and the shear magnitude along each of the $\langle 111\rangle$ slip directions can be assigned uniquely. The displacement gradient matrix can then be unambiguously determined, and the rate of crystal rotation and the rotation axis can be obtained from the curl of the displacement gradient. Standard procedures involved in the calculation of crystal rotations are to be found in the literature (Chin, Nesbitt & Thurston 1966).

As discussed previously, orientations which have zero rate of rotation on straining but which are metastable in the sense that rotation will continue away from the null rotation point if displacements from this position occur, are particularly likely to form transition bands. Some important orientations having zero rotation are to be found among certain families of orientations having $\langle 100\rangle$ or $\langle 110\rangle$ parallel to either the sheet plane normal or the rolling direction, or having $\langle 110\rangle$ or $\langle 111\rangle$ parallel to the transverse direction. Orientations having these zone axes parallel to the physical directions stated rotate about their zone axes during deformation and so the crystal rotations which occur can be shown in two dimensional representation. For all orientations having a $\langle 100\rangle$ transverse direction the rotation rate is zero.

The results of calculations of the rotation rates ($\mathrm{d}\phi/\mathrm{d}\epsilon \equiv$ the rate of rotation in radians per unit natural rolling strain) as a function of orientation in the six families of orientations considered are shown in figure 6*a*–*f*, together with the Taylor factors, M, for the same orientation ranges.

Table 1 is taken from the results of figure 6, and lists all orientations for which $\mathrm{d}\phi/\mathrm{d}\epsilon = 0$. An orientation of zero $\mathrm{d}\phi/\mathrm{d}\epsilon$ may be either stable or metastable with respect to a particular rotation axis. An orientation is considered to be stable relative to rotation about a given axis if when displaced by rotation about that axis the original orientation is restored by the ensuing rotation. In figure 6, the stable orientations are those for which the slope of the curve of $\mathrm{d}\phi/\mathrm{d}\epsilon$ against ϕ is negative at $\mathrm{d}\phi/\mathrm{d}\epsilon = 0$. Metastable orientations are those for which the slope of the curve is positive at this condition, and it is these orientations which may form transition bands by splitting followed by rotations in opposite senses. The stability of the orientations listed in table 1 with respect to displacement about axes parallel to the rolling direction, normal direction and transverse direction respectively are also indicated in the table.

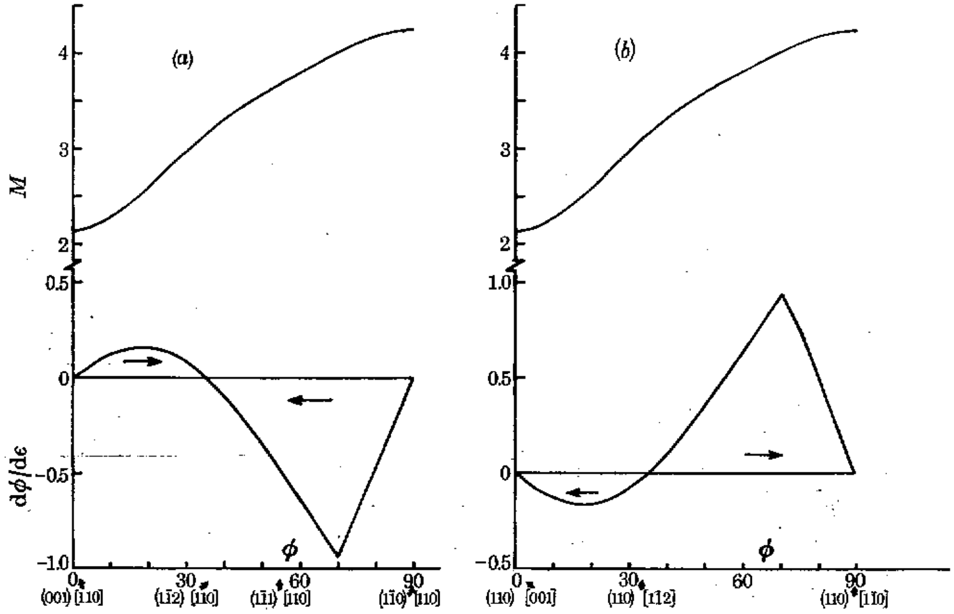
TABLE 1

	rotation axis		
orientation	rolling direction	normal direction	transverse direction
(001) [100]	stable	metastable	stable
(001) [110]	metastable	stable	stable
(110) [001)	metastable	stable	metastable
$(1\bar{1}0)$ [110]	metastable	stable	stable
(110) $[1\bar{1}2]$	metastable	metastable	metastable
$(\bar{1}12)$ [110]	stable	stable	stable
$(\bar{4}, \bar{4}, 11)$ [11, 11, 8]	stable	metastable	metastable
(11, 11, 8) $[\bar{4}, \bar{4}, 11]$	stable	metastable	stable

It may be noted that only {112} ⟨110⟩ orientations are truly stable in that they are self restoring after any displacement. Two other orientations, (001) [110] and (11, 11, 8) $[\bar{4}, \bar{4}, 11]$ (this orientation is near to (111) $[\bar{1}\bar{1}2]$) are shown by results obtained from the rolling of single crystals to be quite stable (Koh & Dunn 1955). The present analysis indicates, however, that (001) $[1\bar{1}0]$ should rotate towards (112) $[1\bar{1}0]$ if displaced about the rolling direction as axis. Computed data for the (11, 11, 8) $[\bar{4}, \bar{4}, 11]$ orientation shows that rotation, initially about the sheet normal, should lead eventually, although very slowly, to (211) $[01\bar{1}]$. The experimental results are not in any disagreement with the present analysis, since the rotation away from the orientations specified would be very slow in each case. It may be noted that results due to Richards & Ormay (1969) show clear indications that rotations from (001) $[1\bar{1}0]$ towards $(\bar{1}12)$ [110] occur in the rolling of polycrystalline iron.

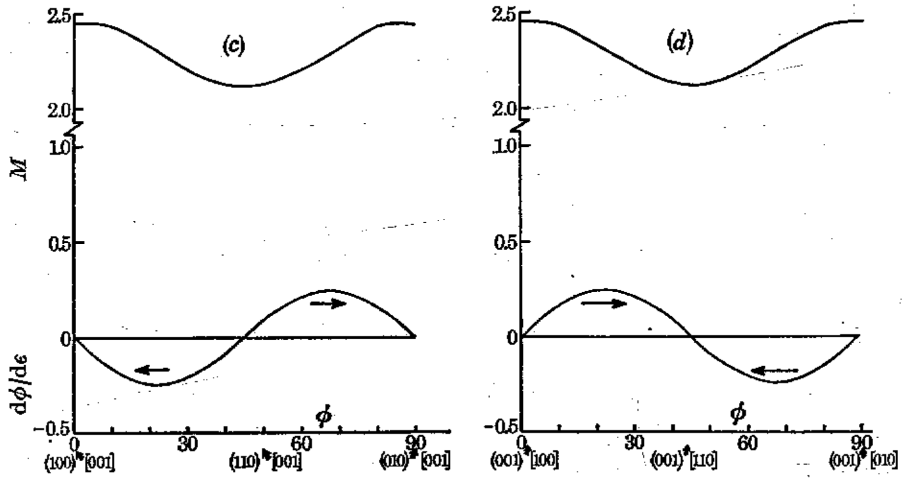
All of the orientations listed in table 1 except $(\bar{1}12)$ [110] are metastable with respect to one or other of the rotations specified and are thus susceptible to transition banding. The band will only form if a curvature is created about the orientation of zero rotation, an angular displacement of the crystal in only one sense merely causes the crystal as a whole to rotate away from the metastable position. Local curvatures inevitably arise as a result of dislocation storage during deformation, and curvatures about the transverse direction as axis inevitably occur in the rolling operation as a result of surface friction effects (Dillamore & Roberts 1963–4). It is thus very probable that a metastable orientation will split into two orientation ranges linked by a transition band. The centre of the transition band will contain a residue of the initial orientation.

Whether the transition band thus formed will act as a centre for the nucleation of recrystallization will depend, as discussed earlier, on the misorientation across the band and upon the structure gradient from the centre to the edge of the band. The misorientation across the band after a given strain depends on three factors. These are the initial curvature (which we may assume to be small and the same in all cases), the strain imposed subsequent to nucleation of the band, and the $d\phi/d\epsilon$ curve for the relevant rotation axis.



[110] parallel to rolling direction

[110] parallel to normal direction

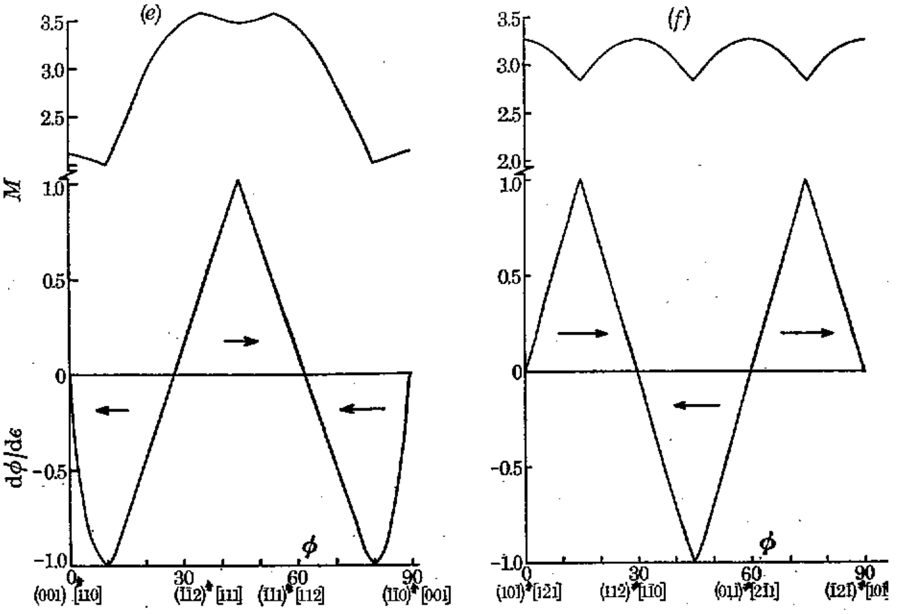


[001] parallel to rolling direction

[001] parallel to normal direction

FIGURE 6. Taylor factors (M) and rotation rates during rolling ($d\phi/d\epsilon$) for five orientation ranges.

The structure gradient will be assumed, on the basis of the results already presented, to be a function of the total shear strain, and this will vary from point to point in the region of lattice curvature as a result of variations in M. A favourable structure gradient for the initiation of recrystallization would have an increasing dislocation density (a smaller average subgrain size and larger sub-boundary misorientation) from the centre of curvature to either side of the band. This requires that the metastable orientation from which the band forms should have a lower M value than the orientations towards which rotations occur.



$[1\bar{1}0]$ parallel to transverse direction

[111] parallel to transverse direction

FIGURE 6 (cont.) For legend see facing page.

From figure 6 it may be seen that M increases on either side of the metastable orientation only for the (110) [001] orientation rotating about both $[1\bar{1}0]$ and [001] and for (001) [110], rotating about [110]. The value of M increases for one direction of rotation only from the $(\bar{4}, \bar{4}, 11)$ [11, 11, 8] with $[1\bar{1}0]$ as axis, and (110) $[1\bar{1}2]$ with [110] as axis. For rotations away from (11, 11, 8) $[\bar{4}, \bar{4}, 11]$, the M value is approximately constant. For the remaining three cases given, M decreases when rotations occur.

The transition band structure gradient, determined by the variation of d_r and σ_r from centre to edge of the band is thus particularly favourable for the (110) [001] orientation and will be favourable also for (001) [110] and rather less so for (11, 11, 8) $[\bar{4}, \bar{4}, 11]$.

The metastable orientations which have decreasing M to either side are unlikely to satisfy the condition represented by equation (1) and will thus not be nucleating orientations. Orientations for which M is high to one side and low to the other represent an intermediate case and may be moderately effective nucleating sites for recrystallization.

There is abundant evidence that (110) [001] is an orientation which readily provides recrystallization nuclei, and Hu (1963) has observed recrystallized grains of this orientation to develop from transition bands. In polycrystalline material it is observed that the (110) [001] orientation occurs most readily on recrystallization in rather coarse grained material which has been deformed by about 50 or 60 %. The grain size dependence of transition band nucleation is expected since the probability of developing the necessary initial curvature is lower with the smaller volume. The strain dependence can be accounted for in terms of the earlier discussion about the importance of the terms θ_t and d_t which define the curvature, and which depend on strain through the relation between $d\phi/d\epsilon$ and ϕ.

If the variation of $d\phi/d\epsilon$ with ϕ is large at the metastable orientation, and if a high rate of rotation is maintained (high $d\phi/d\epsilon$ on either side of the initial orientation) then the curvature in the band will sharpen rapidly with strain. The ensuing structure will be most favourable for nucleating recrystallization at moderate strains, but after large strains will be unfavourable due to too sharp a curvature. This behaviour describes that corresponding to the (110) [001] orientation. A slow rise of $d\phi/d\epsilon$ to a low or moderate value will only give rise to a sufficiently sharp curvature at large strains, and this may correspond to behaviour found in the (001) [110] and (11, 11, 8) $[\bar{4}, \bar{4}, 11]$ orientations. The latter is a commonly observed orientation in recrystallized heavily rolled iron (Haessner & Weik 1956) and may arise by transition band nucleation. The (001) [110] orientation may also be expected to nucleate from grain boundary sites (Dillamore, Smith & Watson 1968), so the importance of transition bands for this orientation is not clear.

A final consideration relates to the incidence of transition bands of different orientations. In a random polycrystal the incidence of grains of any particular orientation is very low and if the metastable orientation never exceeds the random level it is unlikely to provide sufficient nuclei to develop strong polycrystalline textures. However those grains with initial orientations suitable for transition banding may be supplemented by others which rotate to these orientations on straining. Of the three orientations from which transition bands suitable for providing recrystallization nuclei may form, the (110) [001] is likely to be the least abundant since it is fed only by a moderately weak rotation about ⟨110⟩. The (001) [110] and (11, 11, 8) $[\bar{4}, \bar{4}, 11]$ are, however, fed by two strong rotations, as evidenced by the fact that they are well represented in the deformation texture of b.c.c. metals, and will have rather greater probability of providing transition band nuclei for recrystallization.

In commercial production of grain oriented silicon iron, the (110) [001] orientation is increased in volume fraction by a two stage deformation with an intermediate recrystallization anneal. Even so the final anneal produces a rather weak distribution of crystal orientations. The (110) [001] grains are, however, the largest grains in the distribution by virtue of their favoured nucleation, and grow at the expense of the other grains during the secondary recrystallization anneal.

The type of transition banding considered here is, as stated earlier, only one of several types that can arise. Its importance has already been indicated by reference to commercial texture control, but the conclusions about nucleation in regions of high lattice curvature will hold for other cases where the curvature is a result of inhomogeneous deformation.

6. SUMMARY

An analysis of the nucleation of recrystallization in regions of high lattice curvature has shown that:

(1) Intermediate curvatures are necessary; subgrain growth in a region of low curvatures requires a long range to establish a high angle, high mobility interface and in a region of high curvature the sub-boundary angle increases too rapidly for the nucleus to remain viable.

(2) A structure gradient of increasing dislocation density (decreasing d_r and increasing σ_r) on either side of the centre of the orientation spread is to be preferred as providing the highest probability of satisfying the condition for nucleation.

$$D_r \geqslant \frac{4}{3}\left(d_r + d_t\left[\frac{4\sigma_t^2}{\sigma_r^2} - 1\right]^{\frac{1}{2}}\right)$$

(3) The strains at which a band is such as to satisfy these conditions are determined by the relation between $\mathrm{d}\phi/\mathrm{d}\epsilon$ and ϕ.

The region of lattice curvature can form in several ways, but an important example is the transition banding of metastable orientations. Orientations which satisfy the conditions for nucleation of recrystallization in decreasing order of ability are (110) [001], (100) [011], and (11, 11, 8) ($\bar{4}, \bar{4}$, 11]. Other metastable orientations exist but are still less likely to satisfy the necessary conditions.

The results of the present analysis are in agreement with experimental evidence.

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