The Stacking Fault Energy Dependence of the Mechanisms of Deformation in Fcc Metals

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The influence of stacking fault energy on the choice of deformation mechanisms is considered. Experimental evidence is reviewed to distinguish between the possible mechanisms and a limited analysis of crystal rotations is performed. It is shown that twinning, pencil glide, and the more usual octahedral slip are all mechanisms which could be operative under different conditions of stacking fault energy or temperature but that even if only octahedral slip occurs stacking fault energy may affect slip rotations by influencing the choice of operative slip systems. Using the simplifying assumption that deformation is homogeneous, and applying Bishop and Hill's maximum work principle to derive possible combinations of slip systems, a set of criteria are developed to indicate the influence of stacking fault energy on the choice from among the available slip systems, and hence on slip rotations. An empirical method of relating stacking fault energy to texture is outlined and the experimental results obtained by this method are compared with results obtained by other methods.

DEFORMATION textures arise because of the rotation associated with deformation by shear processes, which is so familiar in the case of single and duplex slip observed in single crystal plasticity. If a shear system is defined by shear plane normal h and shear direction u the rotation axis $h \times u$. The rotation thus depends on the choice of shear systems which in turn depends on material and orientation variables. When deformation twinning is an available shear mode there is, of course, an additional reorientation, corresponding to mirror imaging with respect to the twin plane.

Exact theories of deformation texture development cannot be formulated because of the complexity in polycrystalline plasticity, but reasonable approximations corresponding to the upper and lower bound theories of polycrystalline plasticity, due to Taylor[1] and Sachs,[2] respectively, can be obtained. Early theories of the latter type will not be discussed here, since it is considered that the upper bound theory of Taylor is closer to reality. In the upper bound theory it is assumed that each crystal undergoes the same strain as the aggregate as a whole.

In order to impose an arbitrary strain the number of independent variables defining the available deformation mechanisms must be greater than or equal to the number of independent components of the strain tensor. There will in general be a number of variants of a particular shear mechanism available and it is also possible for the deformation to be achieved by mechanisms of a number of different types. In a fcc crystal the selection of independent shear systems will be made from among twelve $\{111\}\langle 112\rangle$ twin systems and twelve $\{111\}\langle 110\rangle$ slip systems. If dislocations can cross-slip freely the slip may be regarded as being on six $\{hhl\}\langle 110\rangle$ systems, this corresponds to the pencil glide case.

The particular combination from among the available modes will be chosen to give the minimum value of M in

$$M = \alpha_i d\gamma_i / d\epsilon \qquad [1]$$

where $d\gamma_i$ is the amount of shear on the i^{th} shear system and α_i is the ratio of the critical resolved shear stress on the i^{th} system to that on the first system; $d\epsilon$ may be taken to be the largest principal strain resulting from shear on the i systems.

In principle α_i could be different for each variant from among the infinity of possible shear modes represented by $\{111\}\langle 112\rangle$ and $\{hhl\}\langle 110\rangle$ and the approach adopted here will be to qualitatively examine the influence of material and process variables on α_i and hence on the choice of shear modes and of shear rotations. First the experimental evidence concerning the operative deformation modes will be reviewed and a later section will consider the rotations to be expected from the different modes. Finally the controlling mechanisms of texture development will be identified and the use of texture measurements as in empirical method of stacking fault energy determination will be described and some of its applications reviewed.

1) EXPERIMENTAL EVIDENCE

It has been recognized for some time that different fcc metals and alloys develop different deformation textures but it was not realized until the work of Smallman[3] that there is a continuous transition in texture which can result from either alloy or temperature variations. Mueller[4] redirected attention to the texture transition which is found in copper rolled at temperatures between 20° and −196°C and Hu and coworkers[5-7] have since shown that the texture transition is found in a wide variety of metals and alloys. Fig. 1 is taken from the work of Hu *et al.*[6] and illustrates the transition in copper. These same workers[5] made the first observations of the texture transition correlation with stacking fault energy; they showed that silver and copper both exhibit a texture transition when rolled at different temperatures, and further, that the texture developed correlated with the stacking fault frequency as determined from X-ray line shift measurements.

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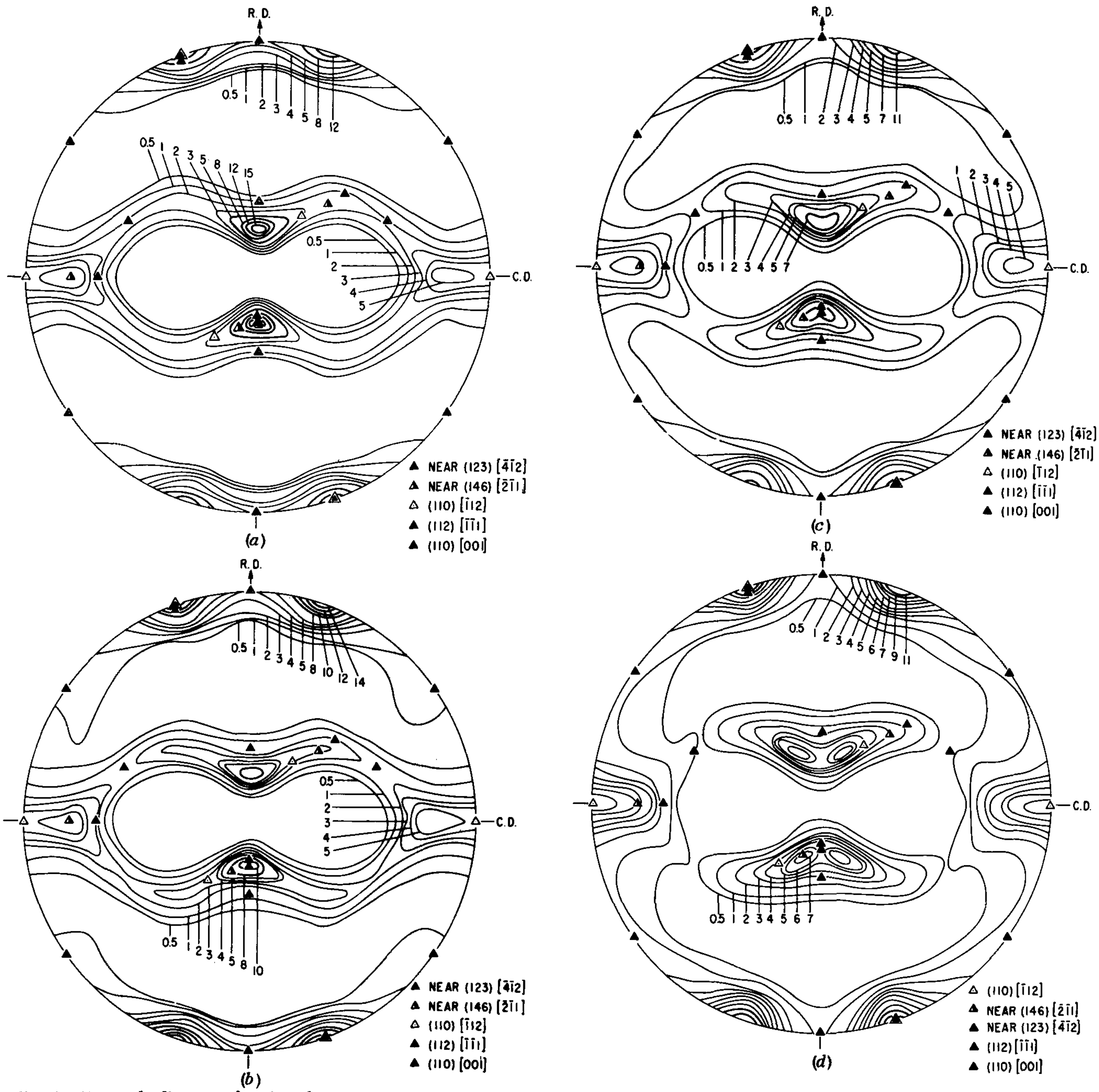

Fig. 1—(111 pole figures showing the transition in texture obtained in electrolytic copper rolled 96.6 pct at (*a*) 25°C, (*b*) −80°C, (*c*) −140°C, and (*d*) −196°C, (after Hu and Goodman[7]).

Smallman and Green[8] and Haessner[9] found a similar correlation for the texture transition which occurs on alloying copper and nickel. This type of experimental correlation led Hu[10] to suggest that stacking fault formation *per se* is responsible for the texture transition, while leading Wassermann[11] to his twinning theory and Dillamore and Roberts[12] to propose cross-slip as a mechanism of importance.

Wassermann and coworkers[13,14] have produced clear pole figure evidence of twin crystals being developed in rolling silver single crystals, thus showing clearly the importance of twinning in low fault energy metals. Hu[10] has produced, in addition to the X-ray line shift evidence, electron micrographs which show either microtwins or bundles of stacking faults in a heavily rolled Cu–4 pct Al alloy, indicating the importance of stacking fault formation. The importance of cross-slip, identified from extensively wavy slip by many workers, has been denied only on the grounds that cross-slip is bound to happen; but, it should be considered here that cross-slip, with the meaning of connecting slip and cross-slip in the sense of pencil glide, are two extremes of a range of possibilities associated with the cross-slip mechanism.

It seems clear that all of the mechanisms advanced have some importance, but does not imply that they are all controlling the texture developed. It remains probable that, in agreement with the postulate of Dillamore and Roberts,[12] cross-slip is the controlling process. An argument along the following lines justifies this viewpoint: when cross-slip is an available means of dislocation escape from stress concentrations the work-hardening rate is reduced, thus making less likely the operation of other mechanisms, such as stacking fault formation and twinning, which require stress concentrations. Stacking fault and microtwin formation, on the other hand sustain the work-hardening rate and would, in fact, thus stimulate cross-slip.

The results of Leffers[15] on the activation energy for the texture transition in a Cu 5 at. pct Zn alloy also support the notion that cross-slip is the controlling process. He found an activation energy of ~10 kcal per mole, which is close to the value found by Wolf[16] for the cross-slip process. Since twinning is generally assumed to be temperature independent, except through the variation in elastic moduli, the correspondence between the texture transition found on lowering the stacking fault energy and that found on lowering the temperature also argues in favor of cross-slip being the controlling process.

Table I. Stress States and the Operative Slip Systems for Axisymmetric Deformation, as in Wire Drawing, Using the Notation of Bishop and Hill[17]

Stress State	Multiplicity*	Equally Stressed Slip Systems	Constricted Systems	Cross-Related Pairs
1	6	a_1 b_1 c_1 d_1 $-a_2$ $-b_2$ $-c_2$ $-d_2$	none	a_1 c_1, b_1 d_1, $-a_2$ $-d_2$, $-b_2$ $-c_2$
4	6	a_2 $-b_2$ c_2 $-d_2$ $-a_3$ b_3 $-c_3$ d_3	$-b_2$ b_3 $-d_2$ d_3	a_2 $-d_2$, $-b_2$ c_2 $-a_3$ b_3, $-c_3$ d_3
7	12	$-a_2$ b_1 c_1 $-d_2$ a_3 $-b_2$ $-c_2$ d_3	$-a_2$ a_3 $-d_2$ d_3	$-a_2$ $-d_2$, $-b_2$ $-c_2$
13	24	a_1 b_1 c_2 $-a_3$ $-b_2$ $-c_3$	a_1 $-a_3$	c_2 $-b_2$
25	8	$-a_1$ c_1 $-d_2$ a_2 $-c_3$ d_3	all	$-a_1$ c_1, a_2 $-d_2$ $-c_3$ d_3

*Under "Multiplicity" is listed the number of stress states crystallographically equivalent to the one indicated in the left hand column.

2) THE SELECTION OF DEFORMATION MECHANISMS

While it may be considered that cross-slip is the process controlling the texture transition, the mechanisms effecting the development of texture remain pencil glide, $\{111\}\langle 110\rangle$ restricted glide, and $\{111\}\langle 112\rangle$ twinning. It is of interest to inquire what crystal rotations result from the operation of these shear modes and thus how material and process variables influence texture development.

Consider first a metal deformed under conditions of temperature and strain rate, and having a stacking fault energy such that both pencil glide and twinning are unlikely to contribute appreciably to deformation, the only important deformation mechanism is then $\{111\}$ $\langle 110\rangle$. Any independent set of five of the twelve $\{111\}$ $\langle 110\rangle$ systems can in principle impose the required shape change, but, using the procedures of Bishop and Hill,[17] the set chosen will be from among the combination of either six or eight equally stressed systems which satisfies the principle of maximum plastic work.[18,19] The systems thus chosen also satisfy Taylor's[1] principle of minimum work, Chin *et al.*[20] having shown identity between the two principles.

Bishop and Hill[17] showed that there are five crystallographically distinct states of stress which stress equally at least five independent slip systems. Symmetry generates fifty-six variants from the five crystallographically distinct stress states. Examples of the five stress states are given in Table I using Bishop and Hill notation, and it will be seen that for each stress state either six or eight systems are equally stressed. It is from among these six or eight systems that the choice of five independent systems must be made.

Plane	(111)			$(\bar{1}\bar{1}1)$			$(\bar{1}11)$			$(1\bar{1}1)$		
Direction	$[01\bar{1}]$	$[\bar{1}01]$	$[1\bar{1}0]$	$[0\bar{1}\bar{1}]$	$[101]$	$[\bar{1}10]$	$[01\bar{1}]$	$[101]$	$[\bar{1}\bar{1}0]$	$[0\bar{1}\bar{1}]$	$[\bar{1}01]$	$[110]$
Notation	a_1	a_2	a_3	b_1	b_2	b_3	c_1	c_2	c_3	d_1	d_2	d_3

It may be noted from Table I that the six or eight slip systems activated by a particular stress state are made up of pairs of slip directions in each of either three or four slip planes. Bearing in mind the dissociation of slip dislocations into Shockley partials and noting that in order to equally stress two $\langle 110\rangle$ directions the maximum shear stress in the $\{111\}$ planes must be along $\langle 121\rangle$ it is apparent that the state of stress will tend either to constrict or extend the slip dislocations.[21] Furthermore inspection shows that dislocations of both Burgers vectors in an active plane are extended or constricted together so that we may refer without confusion to an 'extended plane' or a 'constricted plane'. The effect of the stress states in this respect is shown in column four of Table I where constricted systems are listed; all other systems are extended. Note that there is not complete symmetry between a positive and a negative stress state when the dislocation extension is considered. Reversing the stress state (*e.g.* from 1 to −1) would cause all slip systems to operate in the reverse sense and with the state of constriction reversed. (If, as discussed later, the state of extension influences the choice of slip systems this would lead to a lack of symmetry between the rotations leading, for instance, to tension and to compression textures.)

Slip systems having the same Burgers vector (neglecting sign) are in cross-slip relation to each other and are indicated in the right hand column of Table I.

It has already been observed that an arbitrary strain requires the operation of only five independent slip systems; and there are, for each stress state, a number of separate combinations of five systems which may be chosen from among the equally stressed six or eight systems. For stress states of types 1 and 7 as many as eight combinations may be possible, for type 4 there are generally four, and for types 13 and 25 there are only two combinations of five systems which equally minimize the value of M, Eq. [1]. However, the separate sets of five systems may operate together in any proportions unless some principle can be defined which gives preference to certain combinations as compared with others. The approach to selecting from among the available combination is qualitatively to return to Eq. [1] and to examine whether or not some material variable or some dislocation interraction would raise the value of α corresponding to a particular slip system or combination of slip systems and thus to reduce the prospects of such systems from operating and to favor combinations which minimize their contribution. This approach was first adopted by Bishop[22] who suggested that the phenomenon of overshoot was responsible for textural differences between different fcc metals. He suggested that a tendency to exhibit overshoot would favor the combination(s) of five systems which have the largest individual shear, *i.e.*, which maximize the shear on one slip system. No tendency to overshoot would, according to Bishop, favor combinations in which the total shear ($\Sigma d\gamma = M d\epsilon$) is most evenly distributed between five systems.

Bishop's approach has still some validity but the identification by Hu *et al.*[5-7] of the correlation between stacking fault frequency measurements and the texture developed in different metals, and in the same metal at different temperatures, suggests stacking fault energy as an important variable and stacking fault or twin formation on the one hand or dislocation cross-slip on the other as mechanisms of probable importance. All of these mechanisms involve the state of dissociation of dislocations, which depends on two parameters;[23] γ/Gb and the shear stress on the dislocation (γ is the stacking fault energy, G is the shear modulus, and b the Burgers vector of the unit dislocation).

It follows from this line of approach that the state of extension or constriction of dislocations might be expected to influence the choice from among the available slip systems. It should be noted, however, that this approach cannot be taken too far, different values of α, Eq. [1], for each slip system requires that different states of stress be found from those considered here. Furthermore for very wide dislocations the state of stress may cause the complete separation of the partials, leading to $\{111\}\langle 112\rangle$ slip or twinning and for very narrow dislocations easy cross-glide will allow slip on systems which appear macroscopically as $\{hhl\}$ $\langle 110\rangle$systems; in either case new stress states would apply, not those of Bishop and Hill.

Let us now consider the interaction of the state of stress and material variables with the choice of slip systems, first attempting to find some guidance from the known behavior of dislocations.

Cross-slip provides a mechanism for reducing the rate of work-hardening by allowing dislocations to escape from their slip planes to reduce the back stress on the primary slip system. Cross-slip also provides a means of dislocation multiplication, the Koehler source, generating dislocations on two systems. It may thus be expected that cross-slip will either maximize the shear on a system favored for cross-slip, or maximize the sums of shears on systems in cross-relation which are both, or all, favored for cross-slip. In this context a slip system favored for cross-slip will clearly be one which is constricted by the stress.

Stacking fault formation on a particular slip system probably has the principal effect of making slip on intersecting systems more difficult. Stacking fault formation will thus tend to maximize shears parallel to a particular plane, the plane being one that is extended and thus favored for stacking fault formation. Stacking fault formation would have an effect similar to overshoot but would maximize the shears parallel to a particular extended plane rather than on one system. It may be noted that where there are two or more combinations of five slip systems which equally maximize shears on different extended planes the combinations will be equally probable but mutually exclusive.

The state of constriction and the cross-slip relationships listed in Table I clarify the application of the ideas outlined above. In view of the failure of electron microscopy to reveal the existence of appreciable numbers of Lomer-Cottrell locks it does not appear warranted to develop a detailed theory based on the state of extension of the dislocations and various dislocation interactions; consequently the quite simple approach to the selection of active combinations outlined in Table II has been adopted.

The crystal rotations resulting from any active combination of slip systems can be readily calculated from the curl of the displacement gradient.[22,24] For the general case the rotations occurring require a three dimensional representation except for uniaxial deformation processes or, under lower symmetry deformation, for special families of orientations. Examination of the variation of crystal rotations due to the influence of material variables on the choice of shears will, for clarity, be confined to such special cases.

Rotations occurring for the case of wire drawing under conditions where a) stacking fault formation is favored and b) cross-slip is favored are shown in Fig. 2. It may be seen that $\langle 111\rangle$ and $\langle 100\rangle$ are likely to develop in the ratio $\sim 65:35$ with only slight variation due to the effect of material variables on the choice of shears. The rather small differences predicted in the case of axi-symmetric deformation do not necessarily indicate that the influence of material variables on the choice of $\{111\}\langle 110\rangle$ slip systems has a negligible effect on texture development in the general case. Consider for

Table II

Stress State	Cross-Slip Favored	Stacking Fault Formation Favored
1	Dislocations not constricted. Minimize maximum shear.	Maximize shears parallel to most active slip plane.
−1	Maximize shears on cross-related pairs.	Dislocations constricted. Minimize maximum shear.
±4	Either maximize shears on constricted systems or maximize shears on cross-related pairs.	Maximize shears parallel to most active extended plane.
+7	Maximize shears on $-a_2$ and $-d_2$ together.	Maximize shears parallel to most active extended plane.
−7	Maximize shears on $-b_2$ and $-c_2$ together.	Maximize shears parallel to most active extended plane.
±13	Shear on cross-related systems not variable. Minimize maximum shear.	Maximize shears parallel to most active extended plane.
±25	Maximize shears on cross-related pairs.	Dislocations constricted. Minimize maximum shear.
−25	Dislocations not constricted. Minimize maximum shear.	Shear on individual slip planes not a variable. Minimize maximum shear.
NOTE	It is assumed in numerical analysis that equally probable shear combinations operate equally and together when cross-slip is favored.	It is assumed that equally probable shear combinations are mutually exclusive and give an increased tendency to deformation banding when stacking fault formation is favored.

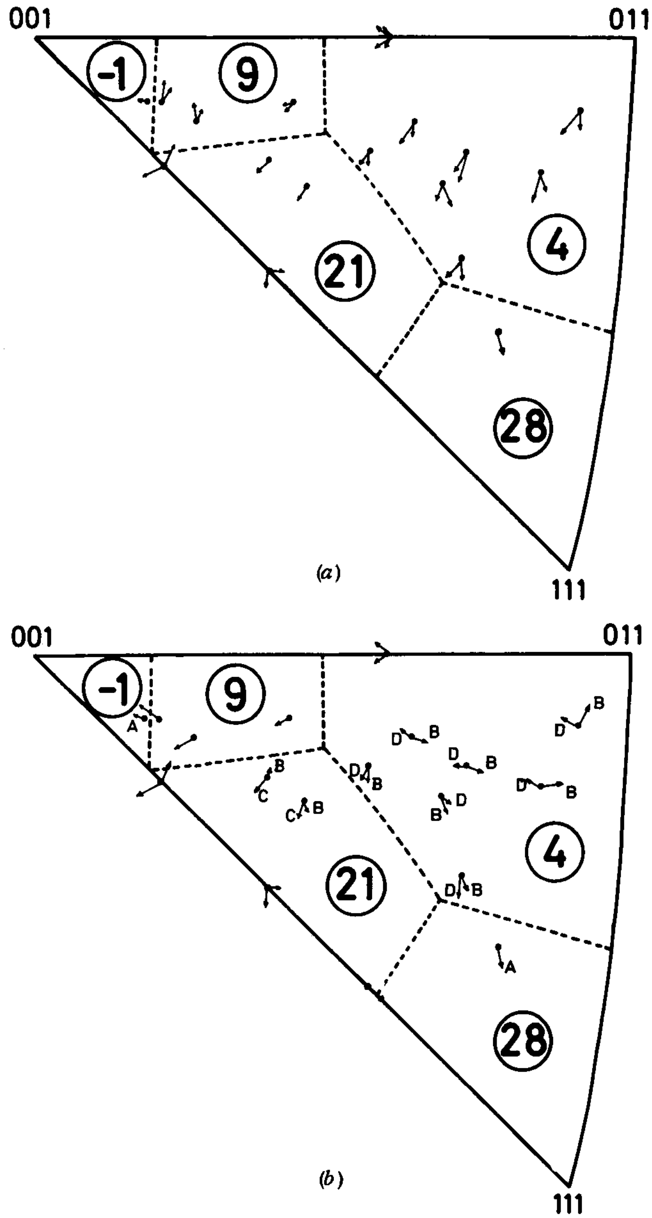

Fig. 2—Fibre axis reorientations predicted in axi-symmetric elongation when (*a*) stacking fault formation is favored (where there is more than one arrow rotations may occur in any direction in the range between the arrows) and (*b*) cross-slip is favored. In the latter case the letter A indicates no ambiguity in choice of rotation, the letter B indicates rotations favored by maximizing the shear on one constricted system, C indicates maximization of shears on two constricted systems which are in cross-slip relationship, and D indicates maximization of shears on cross-related pairs, of which one is constricted, (after Dillamore and Stoloff[21]).

instance the case of deformation by rolling. In Fig. 3 the rotations occurring for crystals having $\{110\}$ sheet planes are indicated for several different circumstances. The rate and sense of rotation are indicated for conditions where stacking fault formation is favored (curve s), where constricted systems are favored (curve c) and where pencil glide occurs (curve p) and for slip on $\{111\}\langle 110\rangle$ assuming that all possible combinations of five systems are equally favored and operate equally (curve a) (note that curves p and a give the same sense of rotation, this is generally true; thus in wire drawing pencil glide would yield the average value for the $\langle 100\rangle:\langle 111\rangle$ ratio). We see from Fig. 3 that when slip is favored on constricted systems less material rotates to (110) $[1\bar{1}2]$ than when slip on an extended plane is preferred.

Except for those orientations which can achieve the imposed strain on less than five independent slip systems (such orientations lie at the boundaries between stress states, where there is no ambiguity in the choice of slip systems) cross-slip and stacking fault formation will favor different crystal rotations. However, since orientation space is divided into cellular volumes corresponding to the regions of activity of particular stress states, the cell boundary network, representing orientations whose sense of rotation is fixed, imposes a basic pattern on slip rotations which does not depend on material variables. The differences in end texture resulting from material differences will thus be more quantitative than qualitative: it is expected that cross-slip being favored would lead to sharper textures.

A limited analysis of 150 randomly chosen orientations indicates that under $\{111\}\langle 110\rangle$ slip the principal orientation spread in the rolling texture would be from $(4,4,11)\,[11,11,\bar{8}]$ to (011) $[21\bar{1}]$ with rather more material towards the (011) $[21\bar{1}]$ end of the spread for materials in which stacking fault formation is favored. This spread is in agreement with that found in 95 pct rolled copper by Bunge and Haessner.[25]

For materials in which the state of dislocation extension is such as to allow of consideration of other shear modes than $\{111\}\langle 110\rangle$, first consider very narrow dislocations which are readily able to transfer, by cross-slip of screw dislocations, from one slip plane to another. Such dislocations will appear macroscopically to slip on a composite plane of the form $\{hhl\}$. The condition for imposing an arbitrary strain is now that

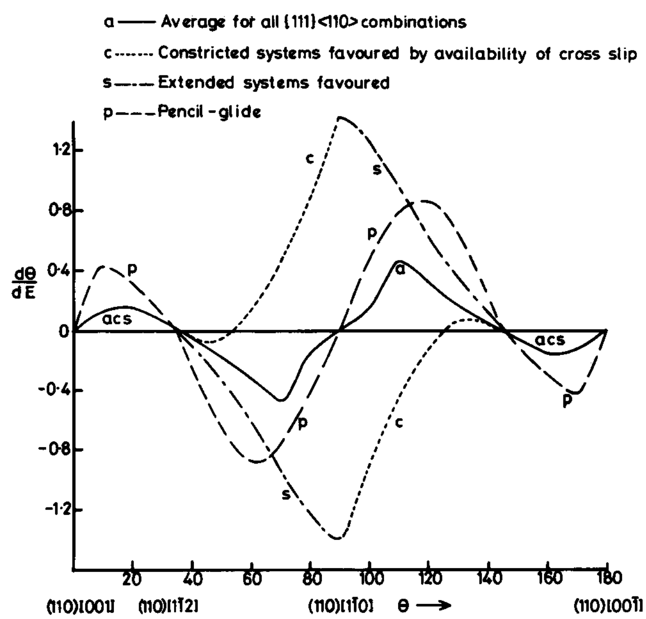

Fig. 3—Slip reorientation rate as a function of orientation for orientation having a (110) sheet plane. $d\theta/dE$ is the instantaneous rate of rotation in radians per unit natural rolling strain. A positive value of $d\theta/dE$ indicates that θ is increased by rotation, a negative value that it is decreased. A zero $d\theta/dE$ value indicates a stable, or metastable, orientation. The four curves are plotted for different mechanistic assumptions.

there should be at least five independent variables. The ⟨110⟩ slip directions are fixed but each slip system is characterized by two independent variables; the amount of shear in each slip direction, and the orientation of the slip planes. Thus, three operating slip vectors will satisfy any imposed strain but it does not follow that only three will operate, since slip on four vectors may require a smaller total shear. Except for certain high symmetry orientations there will be no ambiguity in determining the geometry of deformation although it is generally necessary to use numerical methods of minimizing the shears, there being no simple analytical procedure for the case of fcc pencil glide. The lack of ambiguity in the choice of slip rotations when extensive cross-slip occurs leads to sharper deformation textures being developed, but, as indicated previously the differences will be qualitatively small. A comparison of the rolling textures of aluminum, Fig. 4, and copper indicates the differences to be expected as a result of increasing the stacking fault energy and thus favoring cross-slip, the texture of aluminum being somewhat sharper than that of copper.

For fcc metals of low stacking fault energy twinning is an available deformation mode and Wassermann and coworkers[11,13] have analyzed the reorientation arising from twinning.

It is important to take account of the unidirectional nature of deformation twinning, since this causes certain orientations to be much more likely to twin than others. For orientations near (110) [112] twinning is an unlikely mode since the most highly stressed {111} ⟨112⟩ systems are stressed in the wrong sense. Chin, Hosford, and Mendorf[26] have considered the availability of twinning more generally in relation to the ratio of shear stresses for slip and twinning, and only if this ratio is very high would twinning occur in the (110) $[1\bar{1}2]$ orientation in rolling. On the other hand orientations near (112) $[11\bar{1}]$, including the stable orientation (4,4,11) $[11,11,\bar{8}]$, twin readily, causing twin reorientation to orientations near (001) [110].

It is not possible to analyze twinning according to a homogeneous deformation theory but it is worthwhile to consider how the deviations from homogeneity, which necessarily attend the twinning process are influential in crystal reorientation. Let us suppose that a grain starts to twin. If the whole grain twins a shear strain parallel to the twin system of $\sqrt{2}/2$ occurs within the volume of the grain. The rate of propagation of the twinning dislocation is thought to approximate to the transverse sound wave velocity and the local strain rate will inevitably be higher than the imposed strain rate and the propagating twin, on impinging on the neighboring grain, will cause a stress concentration which can only be relieved by deformation in the surrounding region or by reversed shear, either detwinning or {111} ⟨110⟩ slip within the twin volume. The same will be true even if only part of a grain deforms by twinning, it is not reasonable to view the strain rate as being averaged over the whole grain when only part of the grain twins.

Thus reorientation of a grain due to twinning may be divided into three parts: the twin reorientation of mirror imaging in the twin plane, the shear reorientation due to the twinning, and the slip reorientation due to reversed shear in the body of the twin. Wassermann and Heye[13] have clearly demonstrated that a (112) $[11\bar{1}]$ crystal can reorient in a sequence of twinning and reversed shear to give the (111) $[11\bar{2}]$ component which is so prominent a feature of the low stacking fault energy deformation texture, Fig. 5. It would not be possible to account for this component if reversed shear in the body of the twin were disallowed. This may be seen from Fig. 6 which shows that the (111) $[11\bar{2}]$ orientation is depleted by slip rotations in the sense favored by rolling deformation, a reversed shear could clearly augment orientations near to this one.

Twinning can readily account for the observed differences between the rolling textures of copper and silver which, as pointed out by Wassermann and Heye,[13,14] can be described by the reorientation of orientations near

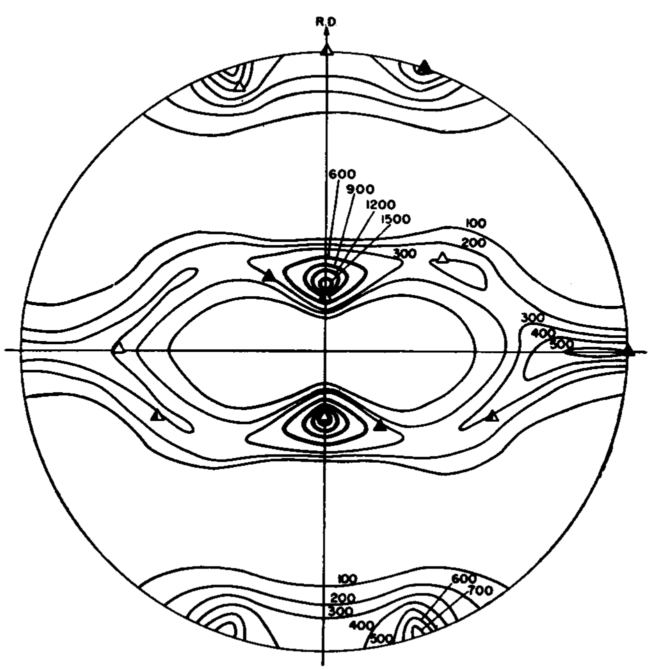

Fig. 4—(111) pole figure for 2S aluminum rolled 95 pct at ~25°C, (after Hu, Sperry, and Beck[39]).

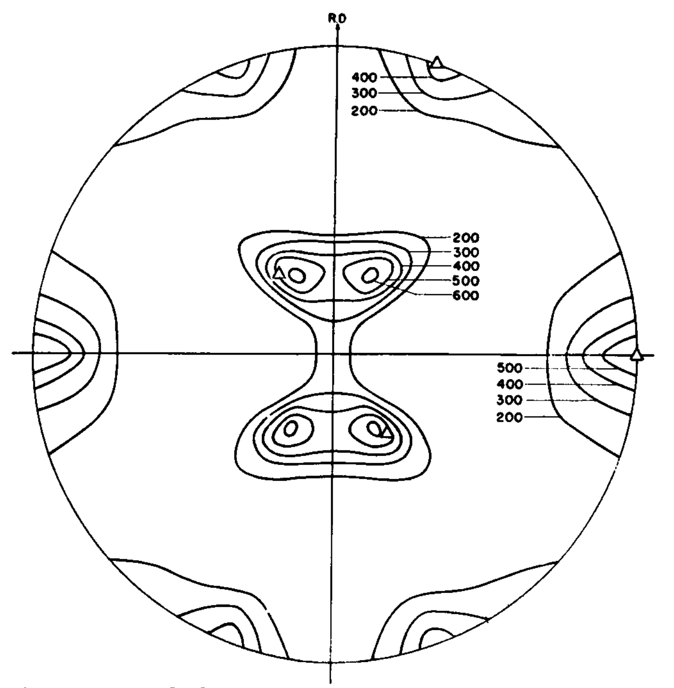

Fig. 5—(111) pole figure of commercial brass rolled 95 pct at ~25°C, (after Hu, Sperry, and Beck[39]).

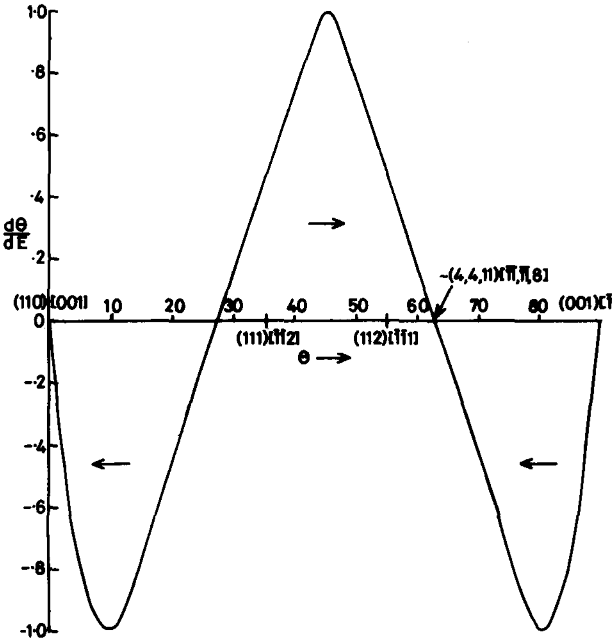

Fig. 6—Slip reorientation rate as a function of orientation for orientations having a [110] transverse direction. Note that there is no dependence of rotation rate on mechanistic variables affecting choice of slip system. The same curve is obtained whether pencil glide or octahedral slip occurs. Note that rotations are away from the metastable region near (111) [$\bar{1}\bar{1}$2].

(112) [11$\bar{1}$] to near (110) [001] and (111) [11$\bar{2}$] orientations.

Current deformation texture theories thus indicate a range of textures depending on the relative contributions to deformation of cross-slip, restricted glide on {111} ⟨110⟩, stacking fault formation, and twinning. In the absence of twinning the texture is expected to consist of a spread of orientations from (4,4,11) [11,11,$\bar{8}$] to (011) [21$\bar{1}$] with rather more material out towards (4, 4, 11) [11, 11, $\bar{8}$] the higher the stacking fault energy, that is, the more cross-slip is favored. When twinning is available the spread about (4, 4, 11) [11, 11, $\bar{8}$] is further diminished by this mode and a component near (111) [112] develops.

In conclusion of this section the fact that deformation does not occur homogeneously must be acknowledged. As shown by Chin[27] the results of English and Chin,[28] Fig. 7, can only be explained if deformation banding is considered, in addition to the processes considered here.

3) THE MEASUREMENT OF STACKING FAULT ENERGY

The conclusion that cross-slip is the controlling process in the texture transition in fcc metals leads to an analogy with the τ_{111} method for determining stacking fault energies. However, the complexities inherent in theories of deformation texture development in polycrystals defy any attempt to produce an exact theory relating texture to fault energy and the method proposed by Dillamore *et al.*[29] is entirely empirical. A pole-figure parameter measured in X-ray transmission geometry is chosen to indicate the relative amounts of materials at the two extremes of the (4, 4, 11) [11, 11, $\bar{8}$] to (011) [21$\bar{1}$] spread referred to earlier. This parameter was chosen to apply to the range of textures obtained from materials in which twinning is not an important mode, and thus to more surely be a function of the ease of cross-slip. It is insensitive to the small texture differences among materials which twin readily.

The procedure in determining, albeit approximately, the stacking fault energy of a sample is the following. A random, fairly fine grained sample is rolled to some standard large strain, in the region of 90 to 99 pct, using as nearly as possible a constant, and standard strain rate. It is not necessary to reverse the sample end to end between passes since this procedure only affects the surface texture, and texture measurements must be carried out on the mid-section.

X-ray intensities are measured in transmission from points at the periphery of the pole figure in the transverse direction (I_{TD}) and at 20 deg to the rolling direction (I_{20}). The ratio I_{TD}/I_{20} characterizes the texture and yields the stacking fault energy of the sample by interpolation on a plot of I_{TD}/I_{20} against γ/Gb obtained by previously following the procedure outlined above for a range of samples of known stacking fault energy. It is, of course, important in comparing textures in this way that they be obtained under equivalent conditions of thermal activation and both the reference and unknown samples are rolled at temperatures which give equal values of kT/Gb^3 in order to approximate as closely as possible to this condition. For most materials the value of kT/Gb^3 can be chosen so that the parameter I_{TD}/I_{20} falls in the sensitive range for the method.

4) SELECTED RESULTS

The principal virtue of the texture method of estimating stacking fault energies is that it can be used for

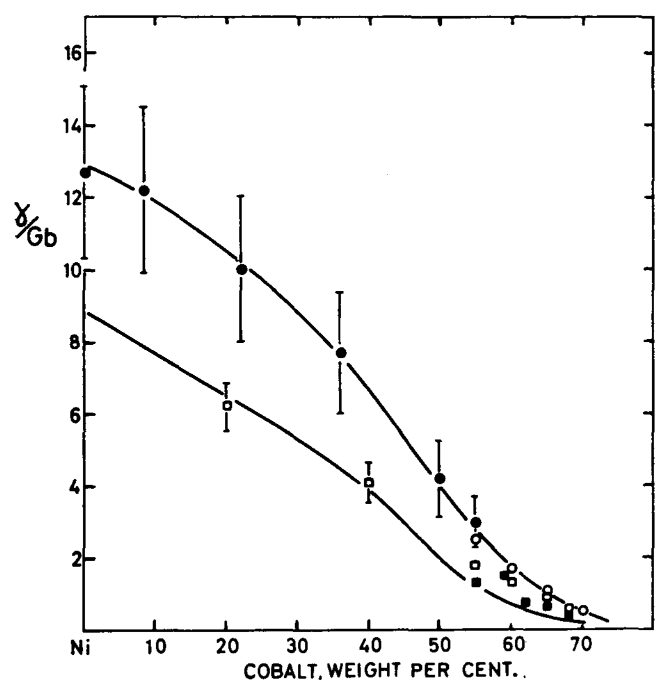

Fig. 7—Stacking fault energy of Ni-Co alloys as a function of cobalt content. Note that the texture results agree well with the extrapolated node values but the stacking fault tetrahedron method gives values at variance with both of these methods. ●—Texture measurements. ○—Node values (Köster *et al.*[34]). ■—Tetrahedron values (Beeston *et al.*[30]). □—Tetrahedron values (Humble *et al.*[40]).

materials in which nodes are not measureably extended, and for alloys in which solute vacancy interactions prevent the use of the climb method. It has been applied principally to investigate the effects of alloying on the stacking fault energy of fcc transition metals.

One way of demonstrating the validity of the method is to standardize against values obtained by a respectable method for a range of materials and then to use the method to reevaluate the fault energy of other materials whose stacking fault energy is known. In Fig. 7 are shown the results of Beeston *et al.*[30] obtained on Ni-Co alloys. The texture method was calibrated against γ/Gb values for aluminum,[31] copper,[32] and silver[33] and it may be seen how well the texture results correlate with the node results of Koster *et al.*[34] for high cobalt content alloys.

A large amount of work has attempted to relate stacking fault energy to valency electron concentration but relatively little attention has been given to the contribution of d electrons to bonding. Perhaps the most unequivocal work on this aspect is that of Harris *et al.*,[35] who used the texture method to determine the stacking fault energy of, among others, the Pd-Ag system, Fig. 8. They showed that below about 60 pct silver the stacking fault energy rises sharply to that of pure palladium, corresponding to increasing the number of holes in the d band. Comparison of the results of Harris *et al.* with more recent results of Rama Rao and Krishna Rao,[36] who used the α-parameter method and report a similar effect, demonstrates the superiority of the texture method in the range of stacking fault energies investigated.

Finally it should be noted that in recent work[37] an attempt has been made to discredit the texture method by applying it on a metal (silver) and at a temperature at which the texture developed is not in the sensitive range of variation. There are, however, other limitations which the texture method shares with most other methods of estimating stacking fault energy; any solute-dislocation interaction may in principle introduce uncertainties. Only one type of solute effect has been observed as clearly interacting with development of texture, namely ordering, either long or short range, but in alloys near to 3:1 or 1:1 atomic fractions. Several indications of the effects of ordering may be found in the work of Beeston *et al.*[30,35,38] and the effect of long range order on texture development has been considered by Dillamore and Stoloff.[21]

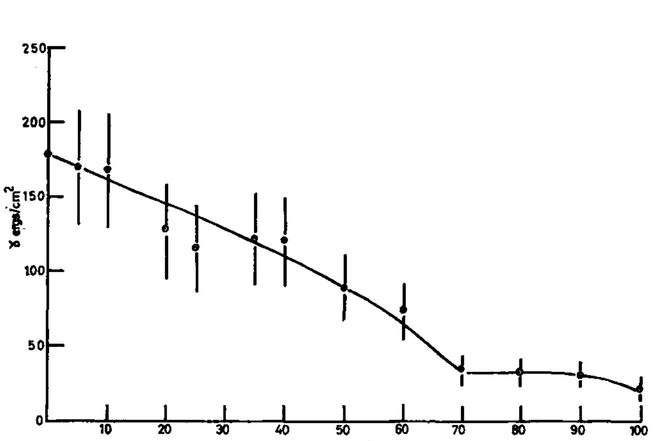

Fig. 8—Stacking fault energy of Pd-Ag alloys as a function of composition.

CONCLUSIONS

The complexity of texture development precludes the formulation of a complete theory defining the effects of material variables. It is, however, apparent that cross-slip is most probably the process controlling the texture transition in fcc metals.

The texture method of estimating stacking fault energy is purely empirical, its principal virtue and justification is that it works and is applicable over the widest range of stacking fault energies, provided that the available thermal energy is adjusted appropriately.

REFERENCES

1. G. I. Taylor: *S. Timoshenko Anniversary Volume,* p. 218, The Macmillan Co., New York, 1938.
2. G. Sachs: *Z. Ver. dent. Ing.,* 1928, vol. 72, p. 734.
3. R. E. Smallman: *J. Inst. Metallk.,* 1955-56, vol. 84, p. 10.
4. H. Mueller: *Oesterr. Akad. Wiss. Math.-Naturw.,* 1958, vol. 7, p. 117.
5. H. Hu, R. S. Cline, and S. R. Goodman: *J. Appl. Phys.,* 1961, vol. 32, p. 1392.
6. H. Hu and S. R. Goodman: *Trans. TMS-AIME,* 1963, vol. 227, p. 1454.
7. H. Hu and S. R. Goodman: *ibid,* 1963, vol. 227, p. 627.
8. R. E. Smallman and D. Green: *Acta Met.,* 1964, vol. 12, p. 145.
9. F. Haessner: *Z. Metallk.,* 1962, vol. 53, p. 403.
10. H. Hu, R. S. Cline, and S. R. Goodman: Recrystallization Grain Growth and Textures, ASM, 1965, Detroit.
11. G. Wassermann: *Z. Metallk.,* 1963, vol. 54, p. 61.
12. I. L. Dillamore and W. T. Roberts: *Acta Met.,* 1964, vol. 12, p. 281.
13. W. Heye and G. Wassermann: *Phys. Stat. Sol.,* 1966, vol. 18, p. K107.
14. W. Heye: Proc. Int. Symp., Textures in Research and in Practice, Clausthal, 1968.
15. T. Leffers: *Scripta Met.,* 1968, vol. 2, p. 447.
16. H. Wolf: *Z. Naturforsch.,* 1960, vol. 15a, p. 180.
17. J. F. W. Bishop and R. Hill: *Phil. Mag.,* 1951, vol. 42, p. 1298.
18. J. F. W. Bishop and R. Hill: *Phil. Mag.,* 1951, vol. 42, p. 414.
19. D. C. Drucker: Proc. 1st U. S. Cong. App. Mech., 1952, p. 487.
20. G. Y. Chin and W. L. Mammel: *Trans. TMS-AIME,* 1967, vol. 239, p. 1400.
21. I. L. Dillamore and N. S. Stoloff: Proc. Int. Symp., Textures in Research and in Practice, Clausthal, 1968.
22. J. F. W. Bishop: *J. Mech. Phys. Sol.,* 1954-55, vol. 3, p. 130.
23. F. R. N. Nabarro: *Phil. Mag.,* 1966, vol. 14, p. 861.
24. I. L. Dillamore, E. Butler, and D. Green: *Met. Sci. J.,* 1968, vol. 2, p. 161.
25. H. J. Bunge and F. Haessner: *J. Appl. Phys.,* 1968, vol. 39, p. 5503.
26. G. Y. Chin, W. F. Hosford, and D. Mendorf: *Proc. Roy. Soc., (London),* 1969, vol. A309, p. 433.
27. G. Y. Chin: Proc. Int. Symp., Textures in Research and in Practice, Clausthal, 1968.
28. A. T. English and G. Y. Chin: *Acta Met.,* 1965, vol. 13, p. 1013.
29. I. L. Dillamore, R. E. Smallman, and W. T. Roberts: *Phil. Mag.,* 1964, vol. 9, p. 517.
30. B. E. P. Beeston, I. L. Dillamore, and R. E. Smallman: *Met. Sci. J.,* 1968, vol. 2, p. 12.
31. P. S. Dobson, P. J. Goodhew, and R. E. Smallman: *Phil. Mag.,* 1967, vol. 16, p. 9.
32. A. Howie and P. R. Swann: *Phil. Mag.,* 1961, vol. 6, p. 1215. L. M. Brown: *ibid,* 1964, vol. 10, p. 441.
33. M. H. Loretto, L. M. Clareborough, and R. L. Segall: *ibid,* 1964, vol. 10, p. 731.
34. E. H. Koster, A. R. Tholen, and A. Howie: *ibid,* 1964, vol. 10, p. 1093.
35. I. R. Harris, I. L. Dillamore, B. E. P. Beeston, and R. E. Smallman: *Phil. Mag.,* 1966, vol. 14, p. 325.
36. P. Rama Rao and K. Kirshna Rao: *J. Appl. Phys.,* 1968, vol. 39, p. 4563.
37. P. C. J. Gallagher and Y. C. Lui: *Acta Met.,* 1969, vol. 17, p. 127.
38. B. E. P. Beeston and L. K. France: *J. Inst. Metals,* 1968, vol. 96, p. 105.
39. H. Hu, P. R. Sperry, and P. A. Beck: *AIME Trans.,* 1952, vol. 194, p. 76.
40. P. Humble, M. H. Loretto, and L. M. Clareborough: *Phil. Mag.,* 1967, vol. 15, p. 297.