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LATTICE-MISFIT THEORY AND TRANSFORMATION TWINNING IN ALKALI FELDSPAR

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ABSTRACT

The geometrical relationships of albite and pericline twins associated with the monoclinic to triclinic phase transformation in alkali feldspar are completely reproduced by lattice-misfit calculations using program EPLAG, and twinned triclinic lattices. Although elastic strain may be an important component in the phase transformation, the irrational orientation of the pericline composition plane is independent of anisotropic elasticity. Elastic strain observed in twin microstructures is due to coherent accommodation of the lattices. These conclusions are consistent with recent studies on oriented phase-boundaries in feldspar intergrowths. The method of calculating elastic strain-energy in EPLAG, which is based on linear strains in the two-dimensional *(hkl)* interface plane, is independently confirmed using tensor shear-strain components calculated from the pericline-twin shear.

Keywords: alkali feldspar, twinning, lattice misfit.

SOMMAIRE

Les relations géométriques des macles albite et péricline qui résultent de la transformation d'un réseau monoclinique à un réseau triclinique dans un feldspath alcalin sont complètement reproduites par calcul de l'écart des réseaux; ce calcul s'effectue au moyen du logiciel EPLAG et en supposant une maille triclinique maclée. Quoique la déformation élastique pourrait être importante dans la transformation de phase, l'orientation irrationnelle du plan de composition de la macle péricline est indépendante de l'anisotropie élastique. La déformation élastique que l'on trouve dans les microstructures maclées est due à la cohérence des réseaux. Ces conclusions concordent avec les études récentes des parois orientées des intercroissances de feldspaths. La façon d'effectuer le calcul de l'énergie de déformation élastique dans EPLAG, fondée sur les déformations linéaires dans le plan bidimensionnel *(hkl)* de l'interface, a été indépendamment confirmée par le calcul des tenseurs de cisaillement pour le déplacement associé à la macle péricline.

(Traduit par la Rédaction)

Mots-clés: feldspath alcalin, macle, écart des réseaux.

INTRODUCTION

Recent studies (Fleet *et al.* 1980, Fleet 1981, 1982, 1984a, b, Fleet & Arima 1985) suggest that the orientation of a phase boundary in intergrowths in which a definite crystallographic relationship exists between the intergrown phases is largely defined by either minimization of lattice misfit or topotaxy. The orientation adopted seemingly provides the optimal degree of fit (or structural continuity) between the juxtaposed substructures of the ligand atoms. Furthermore, whereas many studies clearly show that coherent elastic strain contributes to the total Gibbs free energy of intergrowths and may markedly depress the temperature of phase separation (*e.g.*, Yund & Tullis 1983), the contribution of *anisotropic* elasticity to their orientation appears to be relatively insignificant.

In the present study, the orientations of the composition planes of transformation twins in feldspars are interpreted by lattice-misfit theory. Particular emphasis is given to the pericline twin because of its irrationally oriented twin plane, which suggests some analogy with the coincident lattice-boundaries in intergrowths of chain silicates (Robinson *et al.* 1971, Fleet *et al.* 1980). It is confirmed that the composition planes of both albite and pericline twins are indeed coincident lattice-boundaries and that even though the energy associated with coherent elastic strain may depress the temperature of the monoclinic → triclinic transformation, the orientation of the pericline composition-plane is independent of anisotropic elasticity.

Twinning in inorganic materials and minerals has been reviewed by Cahn (1954). Smith (1974, Chapter 18) gave a comprehensive review of the very extensive literature on feldspar twins and twin-related properties. The origin of polysynthetic lamellar twins in feldspars has been frequently the topic of spirited discussion. However, most of this discussion concerned lamellar twins in plagioclase (*e.g.*, Laves 1965, Vogel & Seifert 1965) which, in light of Smith's (1974) analysis, do not apperar to be related to a phase transformation. Smith (1974, Chapter 18) pointed out that the presence of triplet twin-related reflections in the diffraction pattern of albite- and pericline-twinned plagioclase (T-type twinning, Smith & MacKenzie 1958) excludes an origin by monoclinic → triclinic phase transformation (and also by high triclinic → low triclinic phase transformation). Following the earlier insight of Laves (1950, 1952), the monoclinic → triclinic phase-transformation origin of the fine-scale polysynthetic twins in alkali feldspars is now fairly well established. The present study, then, has application to K-rich feldspar, anorthoclase and cryptoperthite. Relevant studies on

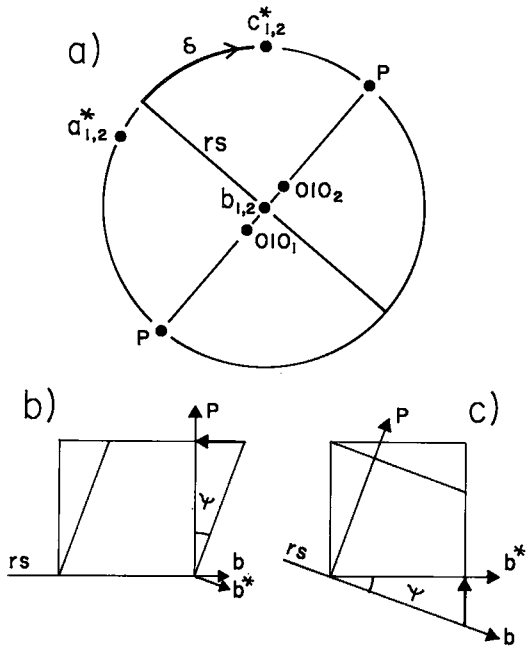

FIG. 1. a) Schematic *b*-axis stereogram of pericline twin geometry: *rs* is trace of rhombic section, with pole *P*; the angle δ is inclination of rhombic section to c^*; subscripts 1,2 label twin-related orientations. b), c) Lattice displacement in the plane of maximum shear-strain (b,b^*), which is normal to the rhombic section, for b) pericline twin and c) albite twin: ψ is the obliquity (or shear angle) of the twin.

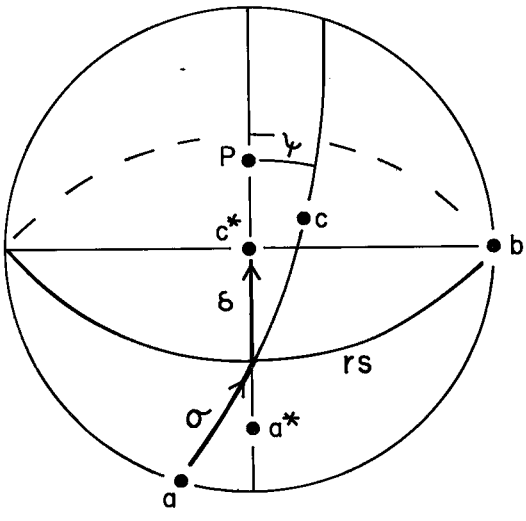

FIG. 2. Orientation of rhombic section *(rs)* defined by the angles σ and δ: see text and Figure 1; stereographic projection.

cryptoperthite include Willaime & Gandais (1972), Willaime *et al.* (1973), Lorimer & Champness (1973), and Willaime *et al.* (1976), and on microcline include Eggleton & Buseck (1980), FitzGerald & McLaren (1982), and McLaren (1984).

TWIN GEOMETRY: REVIEW AND SYNTHESIS

As discussed by Cahn (1954), the albite twin law [twin plane (010)] and pericline twin law (twin axis *b*) are examples of twinning by pseudomerohedry. In the lattice of triclinic feldspar, (010) is a plane of pseudosymmetry and [010] an axis of pseudosymmetry. Thus, for twinning resulting from the monoclinic → triclinic transformation, albite twin individuals are commonly regarded as representing degeneration of the symmetry plane of the pre-existing monoclinic phase. Correspondingly, pericline twin individuals represent degeneration of the symmetry axis.

The orientation of the composition plane of the albite twin is completely defined by the twin law; (010) is both the twin plane and the composition plane. The latter is a coincident plane of the two albite twin-related triclinic lattices. In contrast, the orientation of the composition plane of the pericline twin is only defined by combination of the twin law and the triclinic lattice geometry. The composition plane is the rhombic section (*cf.* Deer *et al.* 1963, pp. 26-29) and, ideally, it too is a concident plane of the two pericline-twin-related triclinic lattices.

In the context of transformation twins, albite and pericline twin geometries are most conveniently described by analysis of the twin shear (*e.g.*, Kelly & Groves 1970, Smith 1974, Fig. 18–20). The albite twin requires a displacement of the pre-existing monoclinic lattice by simple shear parallel to the twin plane in the direction normal to the line of intersection with the rhombic section (Fig. 1c), whereas the pericline twin requires a displacement by simple shear parallel to the rhombic section in the direction of the twin axis (Fig. 1b). For both twins, the plane of maximum shear-strain contains the *b* axis and the (010) normals (this is the b,b^* plane, Fig. 1), and the shear angle ψ is $b \wedge b^*$, which is one-half of the twin shear angle for glide twinning of the triclinic lattice. The angle ψ (Figs. 1,2) is usually referred to as the obliquity of the twin (*e.g.*, Cahn 1954, Willaime & Gandais 1972, Willaime *et al.* 1973). Intersection of the albite and pericline composition planes defines a line of coincidence, which remains invariant in a homogeneous transformation. However, in general, the mutual boundaries of albite- and pericline-twinned regions of complexly twinned intergrowths are not composition planes (Laves 1950).

The rhombic section is defined as the plane through the *b* axis that intersects {010} in a line perpendicular to the *b* axis (Tunell 1952). In stereo-

graphic projection (Fig. 1), its pole P is given by the intersection of the b,b^* and the a^*, c^* great circles. The orientation of the rhombic section is commonly specified by σ, the angle between the a axis and the trace of the rhombic section on the {010} pinacoid (Fig. 2). The angle σ is related to the interaxial angles α, β, γ by:

$$\cot\sigma = \cot\beta - \frac{\cos\alpha}{\cos\gamma\sin\beta} \quad (1)$$

(Tunell 1952). Whereas σ is convenient for morphological study of pericline twin relationships, the true inclination of the rhombic section to the c^* axis (δ, Figs. 1,2) is more appropriate for orientation for the present purposes. Since the b axes of monoclinic and triclinic lattices remain parallel in transformation by pericline twinning, δ is also the true inclination to the c^* axis of the pre-existing monoclinic lattice. Furthermore, it bears a simple relationship to the angular co-ordinate D used in the lattice-misfit program EPLAG (Fleet 1982). By reduction of equation (4) (below), δ is related to the interaxial angles of the triclinic structure by:

$$\cot\delta = -\cos\gamma\tan\omega \quad (2)$$

where,

$$\cos\omega = \frac{(\cos\alpha - \cos\beta\cos\gamma)}{\sin\beta\sin\gamma} \quad (3)$$

Calculation of Lattice Misfit

In the lattice-misfit theory of Fleet (1981, 1982) the orientation of optimal (and coincident) phase-boundaries is predicted by minimizing the area misfit between normalized equivalent *(hkl)* planes of two related lattices. Area misfit is given by $|\epsilon_{11}| + |\epsilon_{22}|$, where ϵ_{ij} is the two-dimensional strain tensor, formed from five linear strains (the sides of a reference triangle, defined by the points of intersection of the *(hkl)* plane with the reference axes, and the bisectors of its two largest enclosed angles). All calculations are made using the program EPLAG.

This theory can be applied directly to modeling the geometry of feldspar twins, using unit-cell parameters for pairs of twin-related triclinic lattices with $a_2 = a_1$, $b_2 = b_1$, $c_2 = c_1$, $\alpha_2 = 180 - \alpha_1$, $\beta_2 = \beta_1$, $\gamma_2 = 180 - \gamma_1$. Since the twin obliquity is only a few degrees, this is the type of problem for which EPLAG is most suited. Poles to computed minima in lattice misfit now indicate the orientation of composition planes. Computed data for lattices of twinned low albite and maximum microcline are given in Figure 3. The original program EPLAG was modified so that the fourth and fifth linear strains for computing ϵ_{ij} relate to the medians through the two largest enclosed angles of the reference triangle. This produces a more equitable distribution of strains for the limiting condition when the reference triangle is almost parallel to a reference axis.

For both low albite and maximum microcline, EPLAG correctly locates one misfit minimum at (010), corresponding to the pole to the composition plane of the albite twin (Figs. 3, 1a), and a second misfit minimum on the primitive circle of the stereogram, corresponding to the pole to the rhombic section, the composition plane of the pericline twin. For the unit-cell parameters of low albite used (Willaime & Brown 1974, Table 1, No. 14), the interpolated δ value is 57°, which compares favorably with the

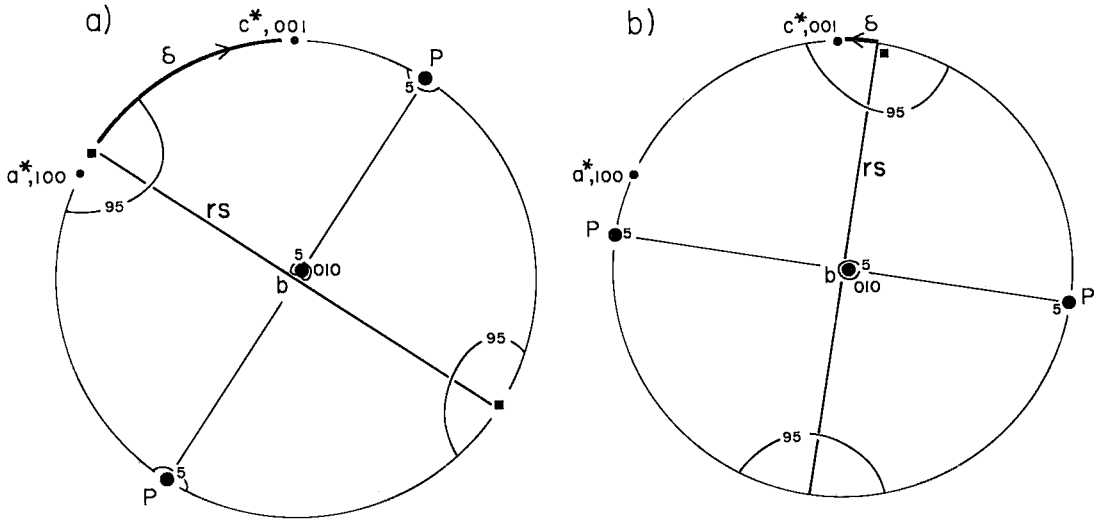

FIG. 3. Lattice-misfit data for a) low albite, b) maximum microcline, *cf.* Figure 1a: large filled circles are minima; squares are maxima; contours are of relative percent lattice misfit; *rs* is rhombic section, with pole *P*; stereographic projection.

precise δ value calculated from equation (2) of 57.3°. For the maximum microcline parameters used (Willaime & Brown 1974, Table 1, No. 9), the EPLAG and precise δ values are -10° and -10.4°, respectively. A single (010) misfit minimum is produced (*cf.* Fig. 1a) because EPLAG essentially computes the misfit of the first lattice relative to the second lattice. The symmetry-related minimum results when the order of the input lattice-parameters is reversed. The area of maximum misfit correctly defines the pole to the plane of maximum shear-strain.

Elastic Strain-Energy

Coherent elastic strain is associated with homogeneous phase separation and phase transformation. The energy required to produce this elastic strain causes a depression of the transformation temperature, as is illustrated schematically in Figure 6 for the monoclinic → triclinic transformation in feldspars. Although Al, Si ordering in feldspars is too sluggish for elegant experimental demonstration of depression of the transformation temperature, there are many observational data to suggest that this transformation is indeed homogeneous (*e.g.*, McConnell 1971, Eggleton & Buseck 1980).

The orientation of the composition plane of the albite twin is defined by symmetry constraints in a homogeneous transformation and is therefore invariant. However, symmetry constraints only define one direction of the pericline composition plane, [010]. On superficial examination, this appears to allow the possibility that the pericline composition plane could be rotated within the [010] zone away from the orientation of lattice coincidence to an orientation that minimizes elastic strain-energy.

Devore (1970) has already noted that there is no correlation between the anisotropy of elastic compliances and the orientation of composition planes of twins in triclinic feldspars. The present study extends this conclusion by showing that the orientation of the pericline composition plane is independent of anisotropy in elastic strain-energy. In general, anisotropy of elastic compliances is not directly equatable with anisotropy of elastic strain-energy. Where there is two-dimensional coherence of phase intergrowths, for example, the dimensional misfit has to be taken into account and, where the misfit anisotropy is large, the misfit completely dominates in the calculation of the elastic strain-energy (Willaime & Brown 1974, Fleet 1982).

The concept of two-dimensional coherence as applied to phase intergrowths (*e.g.*, Cahn 1968, Willaime & Brown 1974) is not strictly applicable to coherent intergrowths of pericline twins because it does not remove the twin shear. With this limitation, the energy associated with two-dimensional coherent elastic strain in the [010] zone is computed here (Fig. 4) as an initial exercise. The calculation of two-dimensional strain-energy uses the misfit tensor (Fleet

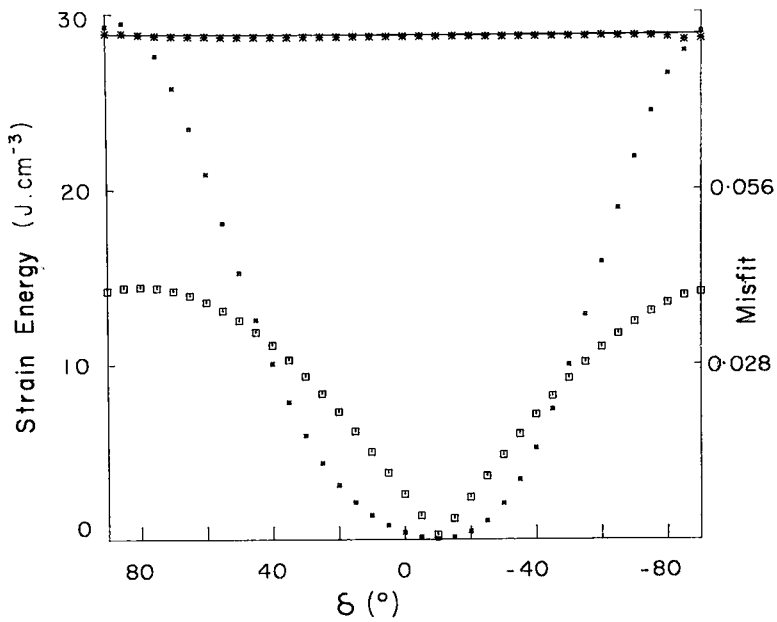

FIG. 4. Lattice misfit and elastic strain-energy for pericline-twinned maximum microcline at various orientations within the [010] zone: open squares are misfit; small asterisks are two-dimensional strain-energy; large asterisks are three-dimensional strain-energy.

1982, 1984a, b) and coefficients of elastic stiffness for microcline (Ryzhova 1964) and albite (Ryzhova & Alexandrov 1965). All calculations of strain energy in this study relate to unit-cell parameters Nos. 14 and 9 (Willaime & Brown 1974, Table 1) for strain-free triclinic lattices of low albite and maximum microcline, respectively. Unit-cell parameters for pericline- and albite-twinned equivalent monoclinic lattices (which correspond to the coherently twinned states) were calculated in the manner of Willaime & Brown (1974). As anticipated, the variation of two-dimensional elastic strain-energy is very similar to the variation in misfit (Fig. 4): the strain energy is zero at the orientation of the rhombic section. It is emphasized that the variation of two-dimensional elastic strain-energy has no bearing on the morphology of pericline-twin intergrowths since the twin shear is not removed in its minimization. It is introduced merely as a preliminary step in the procedure for calculation of three-dimensional data and to provide continuity between the previous applications of program EPLAG to intergrowths of phases (Fleet 1981, 1982, 1984a, b) and the present application to intergrowths of twins.

The three-dimensional coherent elastic strain-energy for the pericline twin, at various orientations of the composition plane within the [010] zone (Fig. 4), has been computed by adding to the two-dimensional misfit tensor, the tensor shear-strain components ϵ_{23}, ϵ_{32}. The latter are given by $\epsilon_{23} = \epsilon_{32} = \{\tan\chi\}/2$, where χ is the shear angle of the twin (ψ) resolved normal to a composition plane in general orientation. The angle χ is calculated with the aid of Figure 5; *bS...* is a composition plane in general orientation within the [010] zone with pole *Pi*. *D* is the angular co-ordinate of program EPLAG (Fleet 1982). Thus χ is defined as $(90\text{–}x)°$. By the methods of spherical trigonometry, x is given from:

$$\cos x = \cos x\cos^2\gamma + \sin x\sin\gamma\cos\gamma\sin D + \frac{\sin x\sin\gamma\cos D}{\tan\omega} \quad (4)$$

where ω is defined by equation (3). Equation (4) can be used also, in a different context, to derive equation (2), which defines the orientation of the rhombic section in the [010] zone, by substituting δ for D and setting $x = 90°$ (Fig. 2).

When *Pi* is coincident with *P* (Fig. 5), $\chi = \psi$ and the composition plane is the rhombic section. Therefore, the two-dimensional misfit-tensor shear components are now zero, and the only non-zero components of the strain tensor are ϵ_{23}, ϵ_{32}. The calculated energies of the coherent elastic strain for the pericline twin with the composition plane in this orientation of lattice coincidence are 63.0 J.cm^{-3} for low albite and 28.7 J.cm^{-3} for maximum microcline. An analogous computation for the albite twin, using a strain tensor formed from ϵ_{23} and ϵ_{32} alone, gives energies of coherent elastic strain of 64.6 J.cm^{-3} for low albite and 28.7 J.cm^{-3} for maximum microcline.

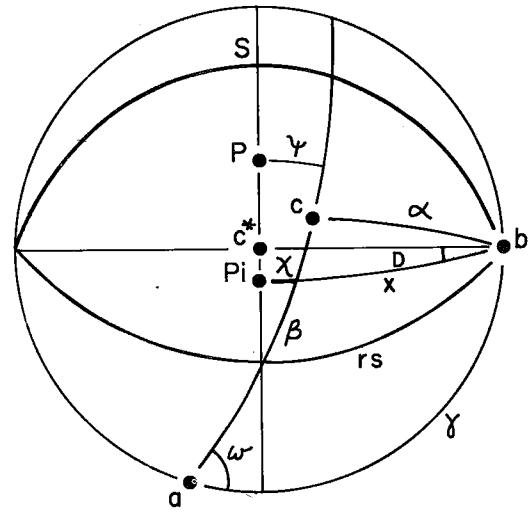

FIG. 5. Definition of χ, the shear angle of the pericline twin resolved normal to a composition plane (*bS...*) in general orientation within the [010] zone, with pole *Pi*: *D* is program EPLAG co-ordinate; *rs* is rhombic section, with pole *P*; stereographic projection.

The data in Figure 4 show that the calculated energies of coherent elastic strain for the pericline twin are essentially independent of orientation of the composition plane. This is readily understood by appreciating that the strain associated with the coherence is defined solely by the displacement vector of the lattice (Fig. 1b). This vector is independent of the orientation of the composition plane and is completely defined by the relaxed (strain-free) triclinic lattice. Albite and pericline twins do have different energies of coherent elastic strain because they have different vectors of lattice displacement; the two vectors have the same magnitude but a different direction (Fig. 1b,c). The small differences in these strain energies therefore reflect elastic anisotropy, largely within the plane of maximum shear-strain (the *b*,*b** plane).

The small discrepancies in the data for three-dimensional coherent elastic strain-energy at various assumed orientations of the composition plane of the pericline twin (Fig. 4) are attributable to use of the misfit tensor in the calculation. This aspect of the present study, in fact, provides a stringent test for the method by which two-dimensional elastic strain-energy is calculated in program EPLAG (Fleet 1982, 1984a, b). Whereas the misfit tensor (which is a measure of area misfit) was considered completely adequate for qualitative use (location of minima,

maxima, *etc.*), its ability to yield fully quantitative strain-energies for simple shear-strained lattices is somewhat surprising. Precise strain-energies at different orientations of the composition plane of the pericline twin are readily computed by forming a strain tensor just from shear-strain components alone (resolved both normal and parallel to the composition plane in general orientation).

Discussion

The present study of pericline- and albite-twinned feldspar lattices has proved to be a stringent test for the methods of calculating lattice misfit and elastic strain-energy using the program EPLAG (Fleet 1981, 1982). The lattice-misfit calculation reproduces the orientation of the pericline composition plane to within $\pm 0.5°$, even with a rather coarse interval of grid points of 5°. The quantitative reliability of the calculated values of elastic strain-energy for simple shear-strained lattices is surprisingly good.

The pericline composition plane is a boundary of lattice coincidence, and its orientation is independent of elastic strain-energy. Although two-dimensional elastic strain-energy is minimized at the orientation of the rhombic section, the twin intergrowth is ideally strain-free in this orientation. Now, the fact that the pericline composition plane is defined by lattice geometry, and is no different from the albite composition plane in this respect, is commonly understood (*e.g.*, Willaime *et al.* 1973, Smith 1974, Chapter 2). What has not been considered in previous studies is the analogy between the irrationally oriented twin-boundary and the oriented phase-boundary.

The coincident two-dimensional phase boundary of chain silicate systems (*e.g.*, Robinson *et al.* 1971, Fleet *et al.* 1980) differs from the irrationally oriented composition plane only in that it relates two different lattices and compositions. Intimate intergrowths of twins commonly exhibit lattice strain (*e.g.*, Eggleton & Buseck 1980), yet the mechanical energy

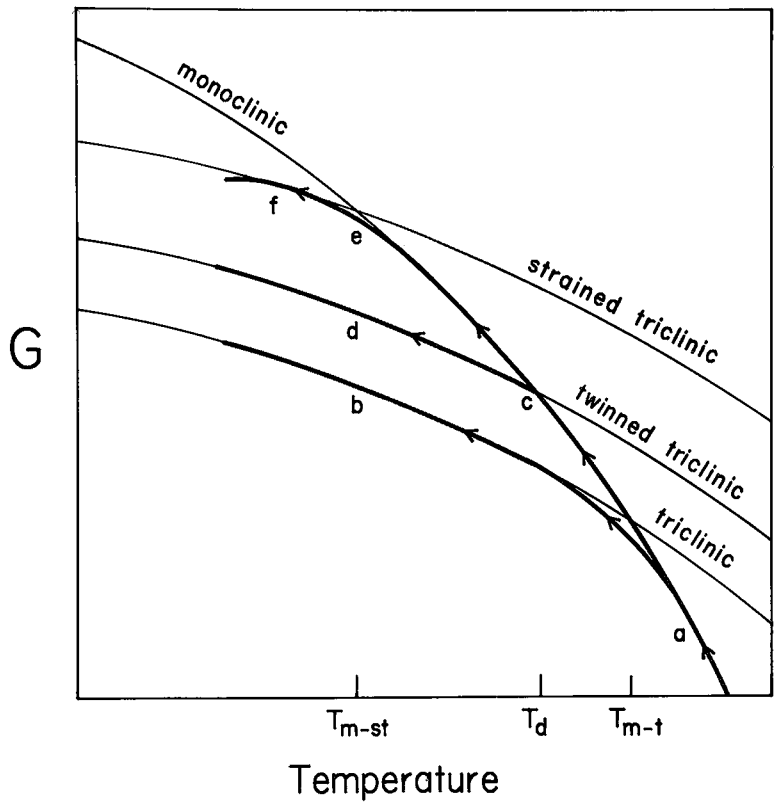

FIG. 6. Schematic Gibbs free energy *(G)* – temperature phase diagram for the monoclinic → triclinic transformation in feldspars: arrows indicate paths of transformation; T_{m-st} is the temperature of the monoclinic → strained triclinic transformation; T_d is the temperature at which disorder is initiated in the triclinic phase; T_{m-t} is the temperature of the monoclinic → triclinic transformation.

associated with this does not contribute to their orientation. The strain in question is readily attributable to twin (or lattice) accommodation. If composition planes are locally rotated away from coincident orientations, compensating counter-rotated composition planes exist elsewhere in the intergrowths. In general, lattice strain *localized* at crystal interfaces is attributable to lattice accommodation (it is imposed by the local structure). Such lattice strain relaxes rapidly with distance away from interfaces. The lack of relevance of the two-dimensional elastic strain-energy of Cahn (1968) and Willaime & Brown (1974) to this situation has been discussed by Fleet (1984a, b). Dislocations at misfit boundaries in framework silicates have a prohibitively high energy because of the high (Al,Si)–O bond strength. Therefore, the common observation of lattice coherence at phase boundaries in cryptoperthites is not at all surprising.

Phase relations for the monoclinic → triclinic transformation in feldspars are presented in Figure 6 in a schematic and greatly simplified manner. A second-order component has been added to the equilibrium transformation path (*ab*..) as required by the Al,Si ordering, but G, T curves are given for just one ordered (or partly ordered) triclinic phase and one monoclinic phase only. Transformation twins in feldspars most likely arise from transformation *via* a supercooled monoclinic state. Under equilibrium conditions, the fine-scale twin lamellae would anneal out because of the large contribution of twin-boundary energy to the free energy. Two extreme paths of transformation are represented by *acef*... and *acd*... (Fig. 6). Many intermediate paths are possible and, indeed, actual transformations may well involve elements of several different mechanisms of transformation. In the first path (*acef*...), Al,Si ordering produces coherent domains of triclinic structure in pericline- and albite-twin-related orientations. Subsequent strain-relaxation and annealing would result in the characteristic *M*-type twinning of microcline. Domain size, shape and distribution could be determined either from a random "nucleation" of regions of Al,Si ordering or completely homogeneously by the establishment of strain waves within the transforming matrix.

An alternative mechanism of transformation is suggested by the path *acd*..., which represents the direct nucleation of an unstrained ordered (or partly ordered) triclinic state from the supercooled monoclinic matrix. The shape and distribution of twin domains must very largely reflect the requirements of strain compensation, twin accommodation and optimization of total energy of the twin boundary. It is important to note that if the transformation takes place below T_d (the temperature at which disorder is initiated in the triclinic phase, Fig. 6), the ordered (or partly ordered) triclinic state may nucleate directly, essentially by a first-order process.

In unannealed feldspars, the orientation of the pericline twin lamellae is a possible fossil record of the structural state of the triclinic phase at the time the twin microstructure was established. This is because the pericline composition plane is defined by Al,Si ordering and, therefore, may not be free to rotate within the [010] zone as the transformation progresses. Of course, if reordering of Al,Si does take place within the time frame of the transformation, this criterion is no longer applicable. One obstacle to its application is the lack of appropriate data on unit-cell parameters for intermediate structural states of microcline and anorthoclase. However, the available data suggest that orientations of the pericline composition plane might vary by only a few degrees in K-rich feldspar. For example, using the intermediate microcline unit-cell parameters of Willaime & Brown (1974, No. 8), δ is $-5.9°$, which is not significantly different from the calculated value of δ for maximum microcline ($\cong -10°$) compared to the observed range of orientations of the pericline composition plane in microcline (Reinhard & Bachlin 1936, Smith 1974, Chap. 18). This one example suggests that for transformation in K-rich feldspar, the variations in the unit-cell parameters of the triclinic feldspar with change in structural state leave the orientation of the lattice-displacement vector essentially constant. In general, the lack of appropriate unit-cell data is a major obstacle to the further application of the lattice-misfit theory to intimate intergrowths of twins and phases and to structural modulations.

Recent TEM studies of transformed alkali feldspars (*e.g.*, Willaime *et al.* 1976, Eggleton & Buseck 1980, McLaren 1984) have revealed much about the nature of their intergrowths but, as reported by McLaren (1984) for cross-hatched microstructure in microcline, the transformation mechanism(s) are still little understood. In view of the numerous physical and chemical factors controlling Al,Si ordering in feldspars (Martin 1974), a full appreciation of their transformation mechanisms may be unrealistic at the present time. The TEM studies do suggest that the transformation path *acef*... combined with random nucleation of Al,Si ordering is unlikely (in twinned microstructures, triclinic regions are twin-pair related, and separate regions of pericline or albite twins exist rather than complex interpenetrating albite and pericline lamellae: Eggleton & Buseck 1980, McLaren 1984). However, in intimately twinned intergrowths in K-rich feldspar, twin (or lattice) accommodation effects lead to coherent and semicoherent structures (McConnell 1971, Eggleton & Buseck 1980). Such studies suggest that the twin microstructure is organized while it develops by strain waves. But this could be interpreted from both homogeneous second-order and strain-compensation

points of view. In the latter case the structure partly relaxes within twin domains as the transformation progresses but remains structurally coherent elsewhere. Transformation paths must be intermediate relative to the two extreme paths in Figure 6. Knowledge of the state of order of structurally coherent and untwinned regions in K-rich feldspar microstructures is critical to a full understanding of the transformation mechanism(s) involved.

Individual Na-rich lamellar regions of cryptoperthites are invariably twinned according to either albite or pericline twin laws (*e.g.*, MacKenzie & Smith 1962, Willaime & Gandais 1972, Willaime *et al.* 1973, Lorimer & Champness 1973, Brown & Willaime 1974, McLaren 1974, Willaime *et al.* 1976, Brown *et al.* 1983, Brown & Parsons 1983). Albite-twinned Na-rich lamellae generally consist of low albite. On the other hand, pericline-twinned Na-rich lamellae within the same hand specimen usually involve high albite (MacKenzie & Smith 1962), although exceptions are not uncommon (*e.g.*, McLaren 1974, Willaime *et al.* 1973). The periodicity of albite-twin lamellae is proportional to the average width of the Na-rich regions and appears to be the result of a delicate balance between strain compensation with the K-rich matrix and minimization of twin-boundary energy (Willaime & Gandais 1972, Brown *et al.* 1983).

The Na-rich and K-rich lamellae are generally coherent along the phase boundary (*e.g.*, Willaime & Gandais 1972, Willaime *et al.* 1976). The required accommodation in the lattice is very localized and the strain relaxes rapidly away from the phase boundary. Now, the lattice-displacement vector for the albite twin is normal to b^* (Fig. 1c) and is always inclined at a low angle to the plane of the lamellae ($60\bar{1}$). The inclination varies from about $+16°$ for anorthoclase to about $-18°$ for low albite. This is very important because it would permit the ordering of the Na-rich phase to develop progressively (continuously or step-wise), without requiring that the composition plane be readjusted with consequent reordering of Al,Si. Relative to the familiar view, in the (001) section of this microstructure, the twin shearing is predominantly downward with only a small component directed toward the K-rich phase.

The lattice-displacement vector for the pericline twin is parallel to b and therefore is parallel to the ($60\bar{1}$) plane for all structural states of albite. However, continued ordering from the high albite state requires rotation of the composition plane of the twin. Thus, the persistence of high albite in pericline-twinned Na-rich lamellae may reflect sluggishness of Al,Si reordering.

M-type twin relationships are expected to be conserved in semicoherent intergrowths. *M*-type twinning certainly indicates that the twin microstructure developed from a monoclinic precursor but it does not preclude the progressive development of low albite.

The present study can shed little light on the control of twin type in individual Na-rich lamellae of cryptoperthites beyond a few obvious statements. Since the total energy (and area) of the twin boundary is optimized, the detailed morphology and orientation of individual lamellar regions have to be important factors. The elastic strain-energies for both twin laws are very similar, but the pericline twin requires a stepped composition surface and therefore would seem to be associated with a higher twin-boundary energy.

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