The preferred orientation in cold-rolled sheet processed from laboratory heats of several titanium binary alloy series was studied. Titanium-based alloys of iron, manganese, chromium, cobalt, nickel, copper, molybdenum, vanadium, columbium, aluminum, tin, and zirconium were included in the investigation. It was found that additions of beta stabilizers caused a texture transition when the retained beta was in excess of about 16 volume percent. This texture transition, for the basal pole, may be roughly described as an unalloyed titanium to zinc-type transition. Copper was an exception and alloys of this element had a tendency to form an "ideal" (0001) [10\bar{1}0] type texture similar to that previously found for additions of the alpha stabilizer aluminum. Neutral alloying elements like tin and zirconium were found to have only a small effect upon the texture, merely sharpening the unalloyed type. Little effect of lattice parameters was noticed except for a trend of high c/a ratio alloys to produce a component of a stable end orientation of the (11\bar{2}0)[10\bar{1}0] type. A qualitative discussion of how these textures form is included; however, due to a lack of complete information about roll-gap stresses and deformation, along with absence of values for critical stresses for slip and twinning in these alloys, it was not possible to be quantitative.
Introduction

It is well known that polycrystalline materials are anisotropic (1). This anisotropy arises from the directional properties of single crystals and is dependent upon the degree of preferred orientation or texture of the polycrystalline aggregate. Many physical and mechanical properties exhibit this anisotropy and it forms the basis for improvement of processing techniques or materials for certain applications. One example of improved processing is that associated with increased drawability for sheet metals (2).

Probably the most successful practical use of anisotropy has been in the field of magnetic properties (3). Technologically, the control of preferred orientation, or texture control, in iron-silicon sheet is further advanced than in any other material. In iron single crystals the cube edge, or [100], direction is known as the easy direction of magnetization, and the [111], or cube diagonal, is the hardest direction of magnetization. Thus in a polycrystalline magnetic sheet material, control of the texture provides a means for improvement of magnetic properties. Recognition of this fact has provided the impetus for much research. Magnetic sheet materials and studies of texture have provided the necessary knowledge, and anisotropic magnetic sheet materials have been produced commercially for several years.

Another potential use of anisotropy is for the improvement of yield strength. Although this has received little commercial attention, it has been the subject of recent research. "Texture hardening"(4) has been demonstrated in the laboratory for plane stress loading of sheet materials. It has been shown that, particularly for tension-tension, the largest improvement in yield strength could probably occur in an hcp metal with an ideal texture (i.e., basal planes oriented parallel to the sheet surface). Obviously, beneficial effects can be achieved in other conditions. For example, a rod in which the c axis is parallel to the rod axis should be 50 to 80% stronger than one with a [1120] fiber texture. Other metals should also show significant increases in strength. For example, a [111] fiber-textured wire should be 20% stronger than a randomly oriented material and 50% stronger than wire with a [100] fiber texture (5). Similar comments can be applied to textured face-centered cubic (fcc) sheet (6).

The potential improvements available through texture control seem to be adequately demonstrated. Why, then, are the practical applications so few? There appear to be many facets to the answer to this question; the most important is that, except for magnetic sheet material and a few cases of deep drawing sheet, industry is not in a position to supply commercial products controlled to a specified texture. The difficulty in obtaining quantitative preferred orientation information has probably impeded progress.
However, improved instrumentation for determining pole figures automatically will alleviate this bottleneck. The demonstration of yield strength improvements through the use of anisotropy has been successful. In most cases, prototypes or laboratory mockups of end items have not illustrated the predicted improvements, thus failing to convince management of the advantages of texture control. These failures can be directly attributed to the lack of specific intense textures in the material employed for the construction of these prototypes. The desired texture simply is not commercially available from the manufacturers. Thus it is imperative to develop substantial technical information so that the required texture control can be achieved, and it is part of the purpose of this program to add to our knowledge on this subject.

There are several mechanisms by which the texture or preferred orientation can be changed. The most important are deformation, deformation temperature, recrystallization, grain growth or secondary recrystallization, and phase transformation. Several complete reviews (7,8) have been published on this subject. Probably the most complete, at least on titanium, is that by Dillamore and Roberts (8). To summarize briefly, considerable study has been made of the deformation texture of heavily rolled sheet material of high purity, somewhat less work on the annealing texture of this material, and some, but very little reported, on the study of the effects of phase transformation of the same material. A limited amount of work has been reported on the effect of alloying elements on the cold-rolled texture of sheet materials in which large changes of preferred orientation were observed, utilizing both pole figure determinations and indirect measurement of mechanical properties. A detailed study of the limited prior work on the effect of alloy elements revealed that a more extensive study could provide a basic experimental program which might yield the information necessary for more fundamental understanding of how textures form. Thus, the main part of this program was to determine the effects of alloying elements on the deformation texture of titanium sheet metal, and to add further to our knowledge relating to texture control.

Experimental Procedure

Alloy Selection

The alloys upon which the program was carried out consisted of several elements of each of the four types (alpha stabilizers, beta isomorphous, beta eutectoid, and neutral). The range of compositions was chosen so as to cover (a) a large range of solubility in the alpha phase (alpha stabilizers and neutrals) and (b) beta stabilizer from pure titanium up to just enough to retain 100% beta (for the beta eutectoid and isomorphous alloys).
Melting, Casting, and Rolling

The alloys were melted in water-cooled crucibles under a positive pressure of welding-grade argon gas, employing thoriated tungsten electrode and a motor generator for a heat source. The castings were thoroughly melted once, then turned over and remelted several times to homogenize the buttons completely. The castings measured approximately 2-1/2" x 2-1/4" x 1-1/4" and normally weighed 450 grams.

These castings were heated in a furnace with argon flowing through. They were then upset to 3/4" thick at a variety of temperatures depending upon the alloy and prior experience with these compositions. The temperature was then lowered to place the alloy in the two-phase alpha-plus-beta phase field or, for the alpha alloys, in the alpha field, and then hot-rolled to a thickness of 1/4". These specimens were then cooled to room temperature and then cold-rolled in one direction, with intermediate anneals at a temperature which was chosen arbitrarily, until a final thickness of 0.040" was reached. No annealing was done after the final cold-rolling pass, which was about 10% in thickness in most cases. In some instances the sheets could not be reduced to the final size, particularly in the higher alloys. In some cases several attempts were made and in the aluminum series a second set of samples was made at 0.100" thick.

Texture Determination

The preferred orientation developed in each of the alloy sheets that were subjected to the above processing was determined by employing the method of Kula and Lopata (9). Such a method produces a first quadrant pole figure, and the other quadrants can be deduced from symmetry. Pole figures obtained by this method were obtained with isointensity contours 10, 20, 30, ... 80. Some of the data were obtained with an automatic pole figure computer* developed part way through the program. These pole figures are identified by isointensity contour levels 1, 2, 3, ... 8. The basal plane (0002) pole figure was determined for all sheets; the prism (1010) pole figure was determined for most of the sheets. However since very little change occurred in the (1010) pole figures they are not reported herein.

Chemical Analysis

A summary of chemical analysis is given in Table I.

TWINNING AND TEXTURE TRANSITIONS IN TITANIUM SOLID-SOLUTION ALLOYS

Table I. Summary of Chemical Analysis for Titanium Base Binary Alloy Series (WT %)

<table>
<thead>
<tr>
<th>Alloying element</th>
<th>Melt</th>
<th>a</th>
<th>b</th>
<th>c</th>
<th>d</th>
<th>e</th>
<th>f</th>
<th>g</th>
<th>h</th>
<th>i</th>
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</thead>
<tbody>
<tr>
<td>Beta Fe</td>
<td></td>
<td>0.12</td>
<td>0.31</td>
<td>0.60</td>
<td>1.18</td>
<td>2.32</td>
<td>4.26</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Eutectoid Mn</td>
<td></td>
<td>.00</td>
<td>.46</td>
<td>0.73</td>
<td>0.91</td>
<td>1.42</td>
<td>5.88</td>
<td>7.09</td>
<td>7.68</td>
<td></td>
</tr>
<tr>
<td>Stabilizer Cr</td>
<td></td>
<td>.00</td>
<td>.09</td>
<td>0.27</td>
<td>0.51</td>
<td>0.97</td>
<td>1.71</td>
<td>3.43</td>
<td>6.81</td>
<td></td>
</tr>
<tr>
<td>Co</td>
<td></td>
<td>.00</td>
<td>.10</td>
<td>0.34</td>
<td>0.83</td>
<td>3.01</td>
<td>5.6</td>
<td>11.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ni</td>
<td></td>
<td>.00</td>
<td>.03</td>
<td>0.33</td>
<td>0.89</td>
<td>1.29</td>
<td>2.7*</td>
<td>5.6*</td>
<td>11.0*</td>
<td></td>
</tr>
<tr>
<td>Cu</td>
<td></td>
<td>.11</td>
<td>.55</td>
<td>1.61</td>
<td>2.0*</td>
<td>4.0*</td>
<td>8.0*</td>
<td>15.0*</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

| Beta Mo          |      | .00  | .33  | 0.67 | 0.94 | 1.35 | 2.70 | 5.42 | 10.74 |
| somorphous V     |      | .00  | .10  | 0.45 | 1.48 | 1.77 | 3.98 | 5.70 | 7.98 | 15.0* |
| Stabilizer Cb    |      | .225 | .53  | 1.00 | 1.96 | 5.17 | 8.36 | 17.14 | 36.0* |
| Alpha Al         |      | .01  | .27  | 0.46 | 0.77 | 1.47 | 2.71 | 3.92 | 8.0 | 15.0* |
| Neutral Sn       |      | .01  | .09  | 0.48 | 1.46 | 1.97 | 4.00 | 8.0* | 15.0* |
| Zr               |      | .03  | .12  | 0.53 | 1.51 | 3.03 | 6.08 | 9.04 | 20.0* |

*These alloys were melted but could not be finished due to severe cracking during rolling.

Discussion of Experimental Textures

Effect of Beta Eutectoid Stabilizers

The textures obtained in this program are similar in many features to those reported by previous investigators (7,8). With the addition of the beta eutectoid stabilizer type elements, the changes in texture, with the exception of copper, formed a similar pattern. It was not possible to verify this pattern for cobalt and nickel because the higher composition alloys cracked during rolling and thus were not available for texture determination. However, manganese, iron, and chromium showed the same pattern. With zero to intermediate alloy additions the basal pole moves slowly from the unalloyed position toward the transverse direction. Sometimes two components in the texture are observed, one near the unalloyed position and one in the transverse direction. As will be shown later, the peak in the transverse direction, a (1120) [1010] texture, is a stable end orientation. With further increase in alloying addition, the beta phase becomes stable and at approximately 16 to 20% volume fracture beta an abrupt transition in the texture is observed. The new basal pole texture is similar to that of zinc or magnesium (7,8) sheet material. However, titanium still has the [1010] parallel to the rolling direction, whereas in zinc and magnesium the [1120] is parallel to the rolling direction. This behavior is best seen by
examining the manganese series, which was one of the most complete, illustrated in Figure 1. The other series are shown in Figure 2 at alloy compositions above and below the transition.

An exception to the above general behavior was found in the copper alloy series which formed nearly an ideal (0001) [10\overline{1}0] type texture with the addition of 0.55 wt % (see Figure 2). One of the important factors in influencing twinning tendency is the grain size (10). The presence of a finely dispersed second phase within a grain has, in fact, the effect of reducing the "effective" grain size and probably would be influential in this suppression of \{11\overline{2}2\} twinning. Thus, the ideal texture produced in copper alloys above 0.55 wt % may be due to the finely dispersed second phase which is clearly visible in the microstructures. This finely dispersed second phase occurs because of the high solubility of copper in alpha titanium near the eutectoid temperature and the decreasing solubility with decreasing temperature.

Effect of Beta Isomorphous Stabilizers

For alloy additions of molybdenum, vanadium, and columbium, the beta isomorphous stabilizers, texture changes that occurred were similar to those of the beta-stabilized eutectoid alloys described in the previous section (see Figure 3).

Effect of Alpha Stabilizers

Additions of aluminum (Figure 4), an alpha stabilizer, caused the basal pole to move toward the sheet normal, producing an ideal or (0001) [10\overline{1}0] type texture similar to that observed by Sparks et al (11). Comparison of the microstructures for the aluminum and the copper series, which also produced this type of texture, clearly showed a very fine structure which could probably suppress twinning.

Effect of Neutral Elements

Additions of tin and zirconium had only minor effects upon the unalloyed texture. The changes noted were primarily that of intensifying the unalloyed texture (Figure 4).

The Role of Lattice Parameters

Considerable effort was expended in determining lattice parameters, with very little evidence that they play a major role in texture formation. It may be that over the range of compositions and alloy types studied, there was too little change in lattice parameters to cause a pronounced effect on texture. There was slight evidence that the higher molybdenum (largest c/a ratio) and manganese (second largest c/a ratio) alloys had a greater tendency for the basal poles to lie in the transverse direction (i.e., to form a (11\overline{2}0) [10\overline{1}0]
Fig. 1. Textures of Ti-Mn binaries.
Fig. 2. (0002) textures of titanium binaries (beta eutectoid).
Fig. 3. (0002) textures of titanium binaries (beta isomorphous).
Fig. 4. (0002) textures of titanium binaries (alpha and neutral).
type texture). From a theoretical point of view, the increase in basal poles in the transverse direction could be attributed either to increased \{1122\} twinning or to increased duplex \{10\overline{1}0\}<1120> slip.

Theoretical Discussion of the Formation of Deformation Textures

Theories of the development of deformation textures are not too well advanced. This is probably due to the complex nature of polycrystalline deformation. The most complete treatment of the formation of deformation textures is that advanced by Calnan and Clews (12-15). Williams and Eppelsheimer have applied the Calnan-Clew method to titanium and conclude, after several assumptions, that by suitable choice of slip and twinning modes and their relative critical shear stresses, it is possible to develop a satisfactory explanation of deformation textures in tension, compression, and cold rolling. The weakness of this method is the lack of information about critical shear stresses for different slip planes and the assumption that twinning is achieved by a critical shear stress. In spite of all these difficulties, this method does have value because it allows qualitative discussion of how textures develop.

The Calnan-Clew Method

The main feature of the Calnan-Clew method is the determination of slip rotation and twinning reorientation caused by an applied stress system. To determine the general grain rotation behavior a resolved shear stress contour diagram for each slip and twinning system having the highest resolved shear stress ratio is plotted in a unit triangle. It is then assumed that only the slip or twinning system having the highest resolved shear stress ratio will be active. The resolved shear stress ratio is defined as the ratio of the resolved shear stress to the critical shear stress. The grain rotation tendencies for various slip and twinning systems are as follows:

1. Simple slip
   (a) Tension - slip direction rotates toward stress axis
   (b) Compression - slip plane normal rotates toward stress axis

2. Duplex Slip
   (a) Tension - the great circle joining two active slip directions rotates toward the stress axis
   (b) Compression - the great circle joining two slip plane normals rotates toward the stress axis
3. Multiple slip

Rotation occurs until stress axis is located symmetrically

4. Twinning

(a) Stress direction does not coincide with shear direction—no deformation

(b) Stress direction coincides with shear direction—rotations obtained from analysis of twin mode

Thus, it can be seen that in order to determine the grain rotations, it is necessary to know the stress system which causes the deformation.

Roll-Gap Stresses and Deformation

One of the main problems in applying any method of determining rolling textures is that the deformation and stresses are known to vary tremendously throughout the roll gap (16,17). This makes computation of grain rotations an almost impossible task. This complexity can be overcome to some degree by making the gross assumption that sheet rolling can be described as a deformation caused by tension in the rolling direction and compression perpendicular to the rolling plane (16-18). This simple model of roll gap stresses has been used by Calnan and Clews (12-15), Williams and Eppelsheimer (19), and Hobson (20). From this model it is possible to calculate the resolved shear stresses on grains of any orientation within a sheet and to determine which slip or twinning mode has the highest stress. It is then assumed that when this stress exceeds a critical value, yielding will occur along with the grain rotation associated with that mode of deformation. Hobson (20), studying cold rolling of zirconium single crystals, reported a computerized program for carrying out the above calculations. Figure 5 shows the results of one such calculation, assuming all slip and twinning modes yield at the same critical stress and that the tensile stress is equal to the compressive stress.

Grain Restraints

Hobson's work (20) was performed on zirconium, which has similar slip and twinning modes to titanium, and it is implied that polycrystalline material behavior is similar to that of single crystals. It is known that this is not entirely true, for Taylor (21) pointed out that in order to maintain grain-to-grain continuity it was necessary that five independent deformation modes occur simultaneously. This continuity requirement and other local grain-grain interactions will undoubtedly cause major effects upon the stresses and deformation modes. Thus it is convenient to accept from the work of
Calnan and Clews, of Williams and Eppelsheimer, and of Hobson that the main deformation mode is that which has the highest resolved shear stress ratio. Furthermore, this main deformation mode produces the principal changes which occur in the texture, even though other accommodation modes are also active.

Application to Unalloyed Titanium

Starting with a random initial orientation and considering that rolling textures are a combination of tension and compression textures, it is now possible to develop a description of how titanium textures form. First, it can clearly be seen that \(\{10\overline{1}2\}\) and \(\{11\overline{2}1\}\) twinning reorient all the grains which have their basal poles in those areas labeled in Figure 5. This reorientation occurs rapidly with small reductions, because the shear associated with these twinning modes is small. The \(\{10\overline{1}2\}\) twinning rotation (about 85°) is such that the new basal pole orientation would be in or near the area labeled \(\{11\overline{2}2\}\). The \(\{11\overline{2}1\}\) twinning rotation would be such that the basal poles move along a great circle toward the transverse direction, about 35°. This is illustrated in Figure 6. The absence of basal poles near the rolling direction in real pole figures supports the twinning theory.

Since it is well established that titanium pole figures have a strong \([10\overline{1}0]\) component in the rolling direction, it is evident that tension in the rolling direction causes \(\{1\overline{1}2\overline{0}\}\) duplex slip on either \(\{10\overline{1}0\}\) or \(\{10\overline{1}1\}\) planes or both. From the compression stresses, \(\{1\overline{1}2\overline{2}\}\) twinning causes a rotation of the grains which have their basal poles near the sheet normal. This rotation, about 64°, occurs about the rolling direction toward the transverse direction. However, real textures of unalloyed titanium do not have basal pole peaks in the transverse direction, thus an additional rotation back toward the sheet normal is required. The most likely mechanism, considering basic principles, would be basal slip. There is a possibility that \(\{1\overline{1}2\overline{1}\}\) twinning could also cause the rotation back toward the sheet normal, since the stress direction would be correct for second-order \(\{1\overline{1}2\overline{1}\}\) twinning within the \(\{1\overline{1}2\overline{2}\}\) twins. The rotation for this second-order twinning would be about 35° from the transverse direction toward the sheet normal; further rotation by slip would be required in order to orient the basal poles properly. In any event, the position of the basal poles along the transverse direction-sheet normal great circle may be considered as the end result of a dynamic process of \(\{1\overline{1}2\overline{2}\}\) twinning, possibly \(\{11\overline{2}1\}\) twinning and basal slip. Therefore the observed cold-rolled texture of unalloyed titanium of \((0001)[10\overline{1}0]\) rotated about 27° toward the transverse direction can be thought of as a combination of \([10\overline{1}0]\) tension textures and a rotated \([0001]\) compression texture. Since rotation occurs about the \(<10\overline{1}0>\) axis, both are compatible. Some of the principal twinning reorientations are shown in Figure 6.
Fig. 5. Stereogram showing the deformation systems with the largest Schmid factor as a function of basal pole orientation. Calculations based on loading with $\sigma_1 = \sigma_2$ and the assumption that each mode has the same critical stress for activation. Between dashed boundaries more than one mode may operate depending on rotation of the basal pole.

Fig. 6. Rotations of the (0001) pole due to various twin (T) and slip (S) modes.
Modification of Texture by Alloying

Changes in texture caused by alloy additions or variations in processing procedures can be explained by modification of the above mechanisms. For the aluminum and copper series the change is quite clear. The ideal (0001)[1010] texture produced in these alloys seems to be a result of the suppression of {1122} twinning. A decrease in the critical resolved shear stress for basal slip could also be effective in producing this texture. For the alloys where a strong (1120)[1010] component shows in the texture, that is in high molybdenum and intermediate manganese alloys, it is similarly clear that this is caused by an increased {1122} twinning tendency and/or a low critical resolved shear stress for (1010)[1120] slip which produces a stable end texture of this type.

In order to explain the "texture transition" of titanium to magnesium type it is, however, necessary to introduce an additional second-order twinning mode. The texture transition is not a true one, for although the basal pole figures for highly beta alloyed titanium are similar to magnesium, zinc, and cobalt, the titanium has a strong [1010] in the rolling direction whereas magnesium has a strong [1120] in the rolling direction. Thus the main features of the tension texture in titanium still remain after the texture transition.

The texture formed by highly beta alloyed titanium can be produced by a second-order twinning of (1012) type occurring within the {1122} twin. The stress axis is in the favorable direction for this mode of twinning. It is interesting to note that the texture transition only occurs when there is a two-phase structure and the amount of the beta phase is greater than approximately 16 to 20 volume percent. It appears that the deformation of the beta phase, which is softer than the alpha, produces a strain field which is conducive to this twinning mode. The twinning rotation would be about 85° along a great circle, 30° from the sheet normal and passing through the transverse direction.

Conclusions

A program consisting of several series of titanium binary alloys was carried out in order to determine the effect of alloy additions on the cold-rolled sheet texture. Beta stabilizer-eutectoid type alloying elements included iron, manganese, chromium, cobalt, nickel, and copper. Beta stabilizer-isomorphous type elements included molybdenum, vanadium, and columbium. Aluminum represented an alpha stabilizer and tin and zirconium were included as neutral elements. A considerable range of composition was covered for each binary alloy.

It was found that most of the beta stabilizers caused a trans-
ition in the texture from a titanium to magnesium zinc type with increasing alloy additions. For some beta stabilizing elements which caused an increase in the c/a ratio there appeared an intermediate texture of the \((11\overline{2}0)[10\overline{1}0]\) type. This texture transition was not to true zinc type because after the transition, the titanium still had \([10\overline{1}0]\) in the rolling direction, not \([11\overline{2}0]\), which is characteristic of zinc. It is apparent that this transition is aided by the presence of a stable beta phase, since the transition in texture was observed only when the amount of beta exceeded 16 volume percent.

Additions of both aluminum and copper increased the tendency to form an ideal or \((0001)[10\overline{1}0]\) type texture. This was known for aluminum; however, the copper results are new. One possible suggestion for these results is that both copper and aluminum form a finely divided precipitate in the alpha phase, thus suppressing \(\{11\overline{2}2\}\) twinning, which is thought to rotate the ideal titanium texture away from the "ideal" orientation.

Additions of tin and zirconium had little effect upon the texture of titanium, only serving to sharpen the unalloyed texture.

General analysis of these textures in the light of the Calnan and Clews method of texture development was carried out in a qualitative way. It was, however, concluded that a quantitative analysis was not possible now because of the complex state of stress and deformation in the roll gap and a lack of knowledge of the critical stresses for slip and twinning in these alloys.

References


3. Graham, C. D., Jr., General Electric, TIS 64-RL-3752M.


