RELATIONSHIP BETWEEN FLOW STRESS RECOVERY AND DISLOCATION STRUCTURE
IN POLYCRYSTALLINE ALPHA-TITANIUM

by

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Thesis submitted to the Graduate Faculty of the
Virginia Polytechnic Institute
in partial fulfillment for the degree of
MASTER OF SCIENCE
in
Metallurgical Engineering

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April 1969
Blacksburg, Virginia
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ACKNOWLEDGMENTS

The author wishes to express his sincere appreciation to for his constant guidance and counsel throughout the investigation. Appreciation is expressed to graduate students, especially faculty, and staff of the Department of Metals and Ceramic Engineering for their valuable assistance.

He wishes to thank the National Science Foundation for financial support of this investigation.

Special thanks are extended to and for typing the thesis.
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I. INTRODUCTION

When a metal is plastically deformed, it strain hardens. This hardening is due to the increase in dislocation density and the mutual interaction between dislocations. Considerable softening may result when a cold-worked metal is given a recovery treatment. A recovery process may be defined as any process, short of recrystallization, whereby the properties of the work hardened metals tend toward their values in the unstrained state (1).

Flow stress may be defined as the stress which is required to produce plastic flow. By straining the material to a predetermined degree of strain and following it by annealing and restraining, the recovery of flow stress can be calculated. A study of the relation between dislocation structures and the associated flow stress is required for understanding of the mechanisms of recovery. This was the object of the present investigation. Similar studies have been performed for other metals (1-6).

It is possible to create rather stable dislocation structures in metals during recovery treatment which improve the high temperature strength of the material. Many authors (1-6) have studied this possibility in FCC metals. Hexagonal close packed metals, however, have not received much attention. For this reason, the present study was made to determine the relation of flow stress recovery to dislocation structure during recovery of polycrystalline alpha titanium. This choice is
further strengthened by the growing importance of alpha titanium in the aerospace field.
II. REVIEW OF LITERATURE

Deformation characteristics of hexagonal close packed metals have been studied in considerable detail by earlier workers (7-16). It has been shown that alpha titanium deforms by both slip and twinning. Mechanical twinning can occur on \{10\overline{1}2\} or \{1\overline{2}n\} (n = 1, 2, 4) planes (16, 17, 18). Further, an attempt has been made (15) to correlate the distinct differences in the shape of twins with the amount of twinning shear.

Three slip systems have been found to operate in alpha titanium (15). They are (0001) <1\overline{1}0> , \{10\overline{1}0\} <1\overline{1}0> , and \{10\overline{1}1\} <1\overline{1}20>. These slip systems can be referred to as basal slip, first order prismatic slip, and first order pyramidal slip respectively. The critical resolved shear stress is lowest for prismatic slip with \{10\overline{1}1\} highest and (0001) highest (7,11,19). According to Tyson (20), the magnitude of the stacking fault energy, \(\gamma\) of the material determines the operating slip system. Metals of low \(\gamma\) prefer to slip on the basal plane and those with high \(\gamma\) on the prism planes. Anisotropic elasticity considerations have also been applied to study the choice of primary slip systems in close packed hexagonal structures (21). These considerations have been extended to study the stability of dislocations on different planes (22, 23).

Conrad et.al. (24-26), and Evans (14) have also studied the deformation characteristics of titanium at low temperatures. The dislocation mechanisms which may be rate controlling are intersection of forest dislocations, Peierls-Nabarro stress, cross slip, and overcoming of
interstitial impurity atoms. Conrad (24) concludes that the rate controlling mechanism during deformation of polycrystalline specimens is that which controls the velocity of dislocations on the first order prism planes. On the basis of the above, this mechanism appears to be the overcoming of interstitial solute atoms according to the model proposed by Fleischer (27) for rapid solution hardening due to asymmetrical distortions. The following relation

$$\tau \propto \text{(Interstitial content)}^{1/2}$$

was found by Conrad (24). $\tau$ represents the shear stress. Churchman (11) has proposed that an increase in the interstitial content would have a much stronger effect on the stress for slip on $\{10\overline{1}0\}$ planes and $(0001)$ planes than on $\{10\overline{1}1\}$ planes. Levine (12) concluded that if interstitial content exceeds 100 ppm, the interstitials would have considerable influence on prismatic slip.

A yield phenomenon has been observed in titanium by many authors (10,19,24,28,29,30). Churchman (28) observed discontinuous yielding in single crystals of titanium containing 0.1 wt% combined oxygen and nitrogen but not in crystals with 0.01 wt% interstitial impurity. He explains the observed phenomena as a combined effect of geometric softening occurring during kink band formation and the yield point phenomenon. The other authors (10,19,24,30) who have reported yield point phenomena were studying materials with greater than 0.1 wt% interstitial impurity.

Wasilewski (29), who observed Luders band formation, does not expect interstitials to be responsible for the yield point phenomena. He
believes that the networks formed on annealing a heavily cold-worked metal are not swept away by recrystallization, and it is these networks that retard dislocation motion during straining.

Since the critical resolved shear stress for various planes is related to the interstitial content, the flow stress of commercial purity polycrystalline titanium should also be strongly dependent on the impurities present. Each interstitial impurity has a different strengthening effect and is usually expressed in terms of oxygen equivalents. These equivalents in parts per million (ppm) are: 1 ppm O = 1 ppm O equivalent, 1 ppm N = 2 ppm O equivalent, 1 ppm C = 1/2 ppm O equivalent (24). Other factors that affect the yield strength are temperature and strain rate (19).

Brehm and Lehr (10) have reported the presence of dislocation loops on tensile straining. But for a few observations, very little has been reported on the substructure of plastically deformed and recovered titanium using transmission electron microscopy (30).

Recovery of metals after prestraining has been studied by measuring various properties, e.g., electrical resistivity, hardness, stored energy, flow stress, etc... A metal which has been plastically deformed can release its stored energy when annealed, by various mechanisms. Li (31) expresses the view that annihilation of dislocation dipoles can be an important recovery mechanism. Annihilation of dislocation loops have been reported in BCC (32) and FCC (5) metals. Yamane (33) using internal friction measurements studied the recovery of titanium. His
experimental data fitted the Cottrell-Bilby (34) point defect equation. This led him to conclude that interaction of point defects with dislocations, i.e., climb, could be an important recovery mechanism. Seeger's (35) work on the recovery mechanisms in titanium supports such a view.

Bailey and Hirsch (36) correlated the distribution and densities of dislocations with the stored energy measurements in polycrystalline silver. Though considerable energy was released during recovery, no significant change was observed in dislocation distribution. They concluded that dislocations in tangles in silver can lower their energy by slight changes in their alignment to one another.

Another important recovery mechanism could be the rearrangements of dislocations into stress free boundaries. Extensive network formation has been reported to be a prime recovery mechanism in some metals (1,4). Lytton et al. (4) have suggested that this process would require climb, cross slip, and glide. Recent studies on aluminum, which has a high stacking fault energy, support such a view (4,5,37). Further, these networks have been shown to form on the original cell walls (37). Polygonization is a special case of this mechanism requiring slip and possibly climb (38).

It is interesting to note that the activation energies for recovery tend toward, but are generally lower than those for self diffusion (1,4,39,40). Li (31) notes a similar effect. He shows that the activation energy should increase with increasing amounts of recovery. Formation and motion of vacancies may be the rate controlling mechanism if
the magnitude of the activation energy for recovery is nearly the same
as that for self diffusion (41).

Flow stress after recovery is a sensitive parameter to changes
in dislocation structure. This parameter has been used previously by
other authors (1,4,5,30,37,39,41). The Dorn \( \theta \) parameter (41,42) is
useful in recovery studies. It is defined as

\[
\theta = t_1 \exp (-Q/RT_1) = t_2 \exp (-Q/RT_2)
\]  

(1)

where \( t \) is the time, \( Q \) is the activation energy for the process, \( R \) is
the gas constant, and \( T \) is the absolute temperature.

The basic assumption involved in the use of equation (1) is that
the properties and structures for condition 1 and condition 2 are the
same. This assumption has been verified for equal values of recovered
flow stress in studies on aluminum (1,4) and recently on titanium (30),
III. EXPERIMENTAL PROCEDURE

The material used in this study was commercially pure, polycrystalline titanium sheet. After suitable cold rolling, specimens of required dimensions were prepared. Recrystallization of the cold rolled specimens was done in a vacuum furnace. The specimens were strained on an Instron machine followed by various recovery treatments. Dislocation structures were examined by using Siemens Elmiskop IA Electron Microscope. The details of the experimental procedure employed are given below.

A. Material

The metal Ti-35A, was purchased from the Titanium Metals Corporation of America. Two sheets of the following dimensions were obtained: 48 inches by 12 inches by 0.083 inches. These were cut from original 48 inch wide strips so that the 12 inch dimension was the prior rolling direction. The as-received material was in an annealed state with a grain size of about 0.037 mm as revealed by mechanical polishing and etching in a 70% HF, 30% HNO₃ solution (30). Chemical analysis and mechanical properties of the heat from which the sheets were taken are given in Table I.

The sheets were labeled A and B and marked off in strips 11/16 inch wide and 12 inches long. These strips were then labeled numerically starting from one end. The initial width of the strip was chosen such that upon rolling to final dimensions, the strip would be 3/4 inch wide.
TABLE I

Chemical Analysis (weight percent)
and Mechanical Properties

<table>
<thead>
<tr>
<th>Heat No.</th>
<th>Ti</th>
<th>C</th>
<th>Fe</th>
<th>N</th>
<th>H</th>
<th>O</th>
</tr>
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<tbody>
<tr>
<td>D 9313</td>
<td>Remainder</td>
<td>0.027</td>
<td>0.06</td>
<td>0.011</td>
<td>0.004</td>
<td>0.08</td>
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<tr>
<th>Direction</th>
<th>Yield Strength</th>
<th>Tensile Strength</th>
<th>Elongation</th>
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<tr>
<td>Longitudinal</td>
<td>30,100 psi</td>
<td>46,800 psi</td>
<td>44.5%</td>
</tr>
<tr>
<td>Transverse</td>
<td>34,700 psi</td>
<td>49,200 psi</td>
<td>38.0%</td>
</tr>
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</table>
B. Tensile Specimen Preparation

The specimen blank was sheared to 11/16 inch width. The as-received sheets were covered with a tape to protect the surface. After shearing, this tape was removed and the surface cleaned with acetone.

A four high rolling mill was used for cold rolling. By laying out the strips in the 12-inch dimension direction, all subsequent rolling was done in the same direction as the prior rolling. A reduction in thickness of 0.006 inch was done in each pass except for the final three passes. These were carried out with a reduction of 0.005 inch for the first pass and 0.002 inch for the remaining two. In this way, strips of 0.083 inch as-received thickness were cold rolled directly to 0.020 inch thickness without any intermediate heat treatment.

A 0.020 inch thick sample was found to be suitable for use in this study. It was not too thick for electro-thinning for transmission electron microscopy and, at the same time, possessed many grains across its cross section. The latter fact ensures that the measured mechanical properties are representative of bulk material. A typical sample which had been rolled to desired thickness, subsequently annealed, and prepared for metallographic examination revealed an equiaxed grain structure with an average grain size of 0.067 mm. The number of grains across the 0.020 inch cross section with this grain size is about eight. The rolled strips were cut into four-inch lengths and marked with the same numbers as the strips from which they were cut. The blanks were then machined to the shape and dimensions of the tensile specimen shown in Figure 1.
FIGURE 1. DIAGRAM OF SPECIMEN USED FOR TENSILE TESTING.
(DIMENSIONS IN INCHES)
A tensilcut machine was used for this purpose. Five or six blanks could be cut at a time using a coolant spray.

C. Vacuum Furnace

Heat treatment of titanium must be done under high vacuum or protective atmosphere to prevent oxidation and contamination. A vacuum furnace designed and built for this purpose was used in the present study. A schematic diagram of the apparatus is shown in Figure 2. It is capable of maintaining a vacuum of $3 \times 10^{-4}$ Torr or better during annealing treatments.

A titanium boat and cover enclosed the samples. This minimized contamination of the samples during annealing. A thermocouple was attached to the boat during annealing treatments and the assumption was made that the temperature of the boat and the samples was the same.

D. Recrystallization Annealing

The procedure employed for recrystallization annealing of titanium samples was the same as in the previous study (30). When the pressure was about $3 \times 10^{-4}$ Torr, with the boat and the samples in the cold zone, the tube and the furnace were brought to a temperature of 816°C (1500°F). After thorough degassing, the boat was pushed into the hot zone using the push rod. The push rod was withdrawn immediately to avoid burning of the vacuum seal around the valve F (see Fig. 2). Starting from the time the temperature of the boat reached 760°C (1400°F), two hours was allowed
Figure 2. Schematic of vacuum furnace
for annealing. At the end of the period the furnace was rolled back and the samples were allowed to cool, the vacuum released, and the samples removed. Each sample was given another number of the form 2B4; the first number, together with the letter B, indicating the number of the annealing batch, and the second number indicating the position it occupied in the stack during the annealing treatment. An average of three measurements of thickness and width measured to the nearest 0.0001 inch across the gauge section was recorded for each sample.

A Debye-Scherrer pattern of a polished section of recrystallized titanium showed rings of fairly uniform intensity (30). This ensures the fact that heavy cold rolling (reduction of about 75%) and subsequent annealing did not develop a strong preferred orientation.

E. Tensile Prestraining

Prestraining of annealed samples to 10% true strain was done on an Instron testing machine. A 1000 pound tensile load cell was used on 50% of the full scale load. The range of a ten-inch chart was therefore between zero and 500 pounds. Since the area of cross section of the tensile sample was about 0.005 square inches, stresses of about 100,000 psi could be read without the need to change the load scale. Though 0.1 inch on the stress axis of the chart paper represented 5 pounds or 1000 psi, it could be read to the nearest pound or 200 psi. With prior calibration, 0.1 inch on the strain axis represented 0.25% engineering strain, although this could be read to the nearest 0.05% engineering strain. A
constant cross head velocity of 0.05 inches per minute was used, providing an actual strain rate of about 0.03 per minute.

The sample was held by a pin and clevis arrangement. Strain was measured using an Instron strain gauge extensometer, model number C-15-12, having a one-inch gauge length. Strain was recorded through an X-Y chart drive amplifier. Calibration of the load cell using 100 pounds of standard weight was done prior to each session of prestraining or restraining.

All the samples were prestrained at room temperature. The procedure used was the same as in the previous study (30) and was as follows:

1. The sample was mounted in the clevis taking care to avoid bending.
2. It was then preloaded to about 10 pounds or 2000 psi, using the same strain rate as in the test.
3. The extensometer was attached and the amplifier was balanced.
4. The sample was pulled to 10% true strain.
5. The sample was unloaded and removed.

Though the recovery of titanium at room temperature was negligible over a period of two days, all the prestrained samples were given their recovery treatments within eight hours after prestraining. For the shorter time recovery tests, the lag time was less than one hour.
F. **Recovery Treatments**

Prestrained samples were given recovery treatments at 500°C (932°F) and 550°C (1022°F) for times of 0.375, 0.75, 1.5, 3, 6, 12, 24, 48, and 100 hours. The procedure followed in the present study for performing recovery treatments was slightly different than the previous one (30). The present procedure eliminated any temperature overshoot and was performed as follows:

1. The boat containing the samples was placed in the cold zone of the furnace tube as shown in Figure 2.
2. The tube was evacuated to better than $8 \times 10^{-4}$ Torr.
3. The furnace was rolled on to the tube and the tube and the furnace brought to equilibrium at a temperature about 50°C (90°F) above the intended recovery temperature.
4. The tube was degassed so that the vacuum was at least as high as before.
5. Using the push rod assembly, the boat was pushed into the leveled hot zone.
6. The push rod assembly was withdrawn and the valve F was closed.
7. When the temperature of the boat was about 175°C (315°F) less than the intended recovery temperature, the furnace power was cut to the set point corresponding to the recovery temperature.
8. As the boat approached the recovery temperature, the transite end plates could be opened and a jet of air used on the furnace tube to prevent any overshoot. Once the recovery temperature was reached, the furnace and the thermac controller maintained it to well within $\pm 2^\circ$C ($3.6^\circ$F) for the duration of the test; the vacuum being held at $6\times10^{-4}$ Torr or better.

9. The test was run for the desired length of time, making compensation for the slow initial heating rate using the Dorn $\theta$ parameter (39,41,42).

10. After completion of the test, the furnace was pulled back and the tube cooled with a jet of air and an electric fan. No compensation was found to be necessary for recovery during the cooling cycle since the cooling rate was fairly rapid.

11. The vacuum was then broken and the samples removed.

A slight discoloration of the samples occurred during recovery due to an inherent leak around the push rod. As the oxide layer dissolved within a second or two during electropolishing, the layer was considered to be too thin to significantly affect the flow stress.

6. Compensation for Slow Heating Rate

Since about 15 minutes was required to heat the boat and its contents to the intended recovery temperature, an approximate
compensation for this slow rate of heating used. This was done using the Dorn θ parameter (39, 41, 42). The basic principle is to equate the amount of recovery taking place in heating to and cooling from the recovery temperature to the equivalent amount of recovery occurring isothermally at the recovery temperature. Use of the Dorn θ parameter requires knowledge of the value of the activation energy Q for the recovery process.

The procedure used to evaluate Q was similar to the one employed in the earlier study (30). The recovery temperature used in the present study, however, were 500°C (932°F) and 550°C (1022°F) instead of 400°C (752°F) and 450°C (842°F). Before the actual compensations as determined by the Dorn θ parameter were calculated, ten additional minutes were allowed for each recovery treatment of three or more hours duration as an approximate compensation.

If it is assumed that samples which have recovered by the same amount are structurally equivalent, and therefore have the same value of θ (1, 39), the value of activation energy can be calculated. Stress-strain curves are quite sensitive to structural differences and thus, identical curves for samples which have recovered by the same amount would indicate structural equivalence. Figure 3 shows that superposition of stress strain curves is obtained for two samples, one recovered for 24 hours at 500°C (932°F) and the other for 3 hours at 550°C (1022°F). They both have a value of f equal to 0.63. Using equation (1) and solving for Q, we obtain an activation energy of
Figure 3. Superposition of stress-strain curves.
52.5 kcal/mole which is close to the value of 57.6 kcal/mole calculated in the earlier study (30). Using the value of Q, the function \( \exp \left( -\frac{Q}{RT} \right) \) was evaluated for different values of temperature. Since the time versus temperature data for heating and cooling was recorded for each recovery treatment, it was possible to plot \( \exp \left( -\frac{Q}{RT} \right) \) versus time for the cycle. The area under this curve represents the temperature-compensated recovery time, \( \theta \). The actual and equivalent isothermal recovery time illustrated in Figure 4 shows how the compensation was made. The area under the dashed rectangle should equal the area under the curve. In this particular recovery treatment an additional nine minutes should have been allowed to compensate for the slow rate of heating, i.e., a three hour "isothermal" treatment would actually last three hours and nine minutes from time of specimen insertion to furnace removal.

As the heating rates varied slightly for each recovery treatment, individual compensations were made for each batch.

**H. Tensile Restraining**

In order to estimate the recovery in flow stress, the recovered samples were restrained using the same procedure outlined above for prestraining.
Figure 4. Typical variation of $\exp(-Q/RT)$ with time
I. Preparation of Thin Foils from Bulk Samples

In the earlier study on alpha titanium, some artifacts were observed during transmission electron microscope studies of thin foils (30). To check the possibility of any aging or contamination occurring during recovery annealing, an additional annealing treatment of 100 hours at 400°C was given to a few samples from the annealed stock. Stress strain curves after this treatment showed no strengthening and the foils seen by transmission electron microscopy showed no precipitates (30). Except for a few hydride patelets formed during electropolishing, no precipitates were observed in the present study.

The thinning technique employed in the present study was quite different from the one employed previously (30). Chemical thinning of titanium at room temperature (10,43) and at lower temperatures (30) was tried. In the foils obtained by this method, it was difficult to find uniformly thin regions suitable for transmission electron microscopy.

Electropolishing of titanium has been used by various groups (44, 45). In the present study, electropolishing of titanium yielded thin foils which had large areas suitable for transmission electron microscopy. The procedure followed was as follows:

1. The 0.020 inch thick gauge section was cut from the tensile sample using a power shear.

2. The electropolishing solution (44,45) used consisted of 60% methanol, 34% n-butanol, 6% perchloric acid (all by
volume). About 200 cc's of the solution was used to keep the sample submerged in the solution during electropolishing. The solution was maintained at a temperature of \(-45^\circ\text{C} (-49^\circ\text{F})\) or lower. This was made possible by surrounding the beaker containing the solution with ethyl alcohol and dry ice. Bits of dry ice were added at frequent intervals to maintain the temperature at the required level.

3. The top edge of the sample was coated with Lacomit stop-off lacquer so that at least one inch of the sample remained to be thinned. The sample was introduced into the solution and electropolishing carried out at 15 volts and 0.2 to 0.3 amps using a circular platinum cathode.

4. Electropolishing was continued for about 1-1/2 to 2 hours with the sample being continuously stirred with horizontal and vertical motions.

5. After this period, the sample was removed and quickly introduced into a beaker containing methanol maintained at 0°C. The sample, after washing in methanol for about two minutes, was introduced into two other beakers containing methanol. These solutions were also maintained at 0°C. A thorough washing with methanol for about 5 minutes prevented surface contamination. The sample was dried between two pieces of filter paper with a warm blast of air.
6. The sides of the sample and the bottom edge were coated with Lacomit and thinning continued until the appearance of a hole in the sample. At this stage, step 5 was repeated and a small area adjacent to the hole covered with Lacomit.

7. Final thinning was carried out at a temperature of about -50°C (-58°F) with continuous stirring of the sample. The rate of thinning was very slow at this stage.

8. With each appearance of a hole in the sample, step 5 was repeated and Lacomit applied to the hole. When a hole appeared near the center of the specimen, the specimen was washed thoroughly with methanol and then introduced into a series of beakers (usually three to four beakers) containing acetone maintained at 0°C to dissolve the Lacomit completely. After drying the sample, the area adjacent to the center hole was cut carefully with a scalpel. It was found that by the method of stirring employed, the thin foil had large thin areas suitable for transmission electron microscopy. The rest of the sample was washed and dried thoroughly after electropolishing for about five seconds. Some more thin foils were cut with a scalpel.

Generally, the above procedure required about six to eight hours for preparation of a thin foil from the original specimen.
J. Transmission Electron Microscopy

A Siemens Elmiskop IA microscope was used at an accelerating voltage of 100 kilovolts to study the dislocation structure of alpha titanium. A Swann double-axis tilting stage was used to tilt the foil. Standard Siemens instructions were followed for alignment and operation. Samples examined by transmission electron microscopy were chosen after the $f_x$ versus time curve had been determined (Figure 5). The samples examined were: annealed material, as-prestrained material, and those given recovery treatments for 0.375 hours, 6 hours, and 100 hours at 500°F (932°C), and for 6 hours and 100 hours at 550°F (1022°C).

In all cases, three to four foils were examined. The general practice was to take micrographs which were typical of the sample. In some foils it was found that precipitates, probably titanium hydride, were found near the thin regions of the foil. The average foil thickness, estimated by counting the number of fringes, suitable to take micrographs was about 4000 Å.

K. Analysis of Selected Area Diffraction Patterns

Solution of selected area diffraction patterns of HCP metals is much more complicated than for cubic crystals. Useful formulas for analysis of SAD patterns were obtained for the previous study (30) from Otte and Crocker (46), Rarey et al. (47), Hirsch et al. (48), and Partridge and Gardner (49). Unfortunately, some of the formulas given in the first two papers contain mistakes. Corrected formulas (30) are
given in Appendix, Table I. In the HCP patterns given by Hirsch et al.,
the three axis notation is used instead of the four axis-four index no-
tation or the four axis-three index notation that is most commonly used.

Using the formulas given in Appendix, Table I, together with a
ratio chart and tables of interplanar angles (50,51), the SAD patterns
were solved. A Kikuchi map for HCP crystals (52) was also found to be
a useful guide in the analysis. In all, about ten different foil orien-
tations were found, indicating that the material did not contain a high
degree of preferred orientation.
IV. RESULTS AND DISCUSSION

A. Flow Stress Recovery

The fractional flow stress recovery parameter, \( f_r \), is defined as the decrease in flow stress due to recovery treatment divided by the decrease experienced if the material recovered completely to the unstrained state \((1,4)\). Mathematically, it is defined as shown in Figure 6. The value of yield stress, \( \sigma_y \), showed a slight variation from sample to sample. However, the average difference between \( \sigma_y \) and the stress corresponding to 10% true strain, \( \sigma_1 \), remained nearly constant.

Alpha titanium, like other metals such as single crystals of iron, cadmium and zinc, exhibits yield point behavior. This heterogeneous deformation introduced the problem of determining the initial flow stress appropriate for uniform deformation. The value of \( \sigma_y \) was therefore read with a 0.2% offset from the elastic loading line by extrapolating the uniform hardening curve as shown in Figure 6.

A plot of the fractional flow stress recovery parameter versus the logarithm of recovery time is as shown in Figure 5. The variation of \( f_r \) with recovery time will be discussed shortly. Stress-strain curves for samples pretrained at room temperature and subsequently restrained after recovery for various times are shown in Figures 7 and 8. A complete tabulation of the stress strain data and the values of \( f_r \) is given in the Appendix tables II and III.

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Figure 5. Fractional flow stress recovery of alpha titanium as a function of recovery time at 500°C and 550°C. The numbers in parenthesis indicate the number of tests performed to obtain the average value shown. The dashed curve is the data of a co-worker(30).
Figure 6. Illustration of the method used for calculating fractional flow stress recovery parameter $f_r$. 

\[ f_r = \frac{\sigma_1 - \sigma_2}{\sigma_1 - \sigma_y} \]
Figure 7. Room temperature stress-strain curves of alpha-titanium for various recovery times at 500°C (932°F).
Figure 8. Room temperature stress-strain curves of alpha-titanium for various recovery times at 550°C (1022°F).
The average values of $\sigma_y$ and $\sigma_1$ were 22,372 psi and 42,538 psi respectively. All the $\sigma_1$ values were within 5% or 2,100 psi of the average value and for $\sigma_y$ all were within 10% or 2,200 psi of the average value. The value of $\sigma_y$ indicated in the supplier's analysis was about 8,000 psi higher than the average value reported above. Though the reason for this is not fully understood, it may be partially attributed to their use of a higher strain rate than that used in the present study. In addition, the grain size of the specimen tested by the supplier may be smaller than that employed here.

B. Yield Point Phenomenon

In a majority of the prestrained samples, yield point behavior was observed. This phenomenon can be attributed either to Churchman's interstitial locking mechanism (28) or Wasilewski's network mechanism (29). As reported earlier, Churchman (28) observed yield point phenomenon in single crystals of titanium containing 0.1 wt% interstitials but not in crystals containing 0.01 wt% interstitials. Wasilewski (29) holds the view that dislocation networks formed on annealing a cold-worked metal are not swept away by recrystallization. Upon straining, these networks impede dislocation motion. Thin foils prepared from recrystallized specimens, however, did not show the presence of any networks. It is quite unlikely that they were lost during thinning, since networks formed during recovery treatments were not lost. For a discussion about the possible loss of dislocations and rearrangements in structure
during thinning, refer to Lytton et al. (4). The material used in the present investigation has an interstitial content of more than 0.1 wt%, and it is likely that interstitial locking of dislocations is the cause for the observed phenomena.

C. Mechanism of Plastic Deformation

When slip occurs on multiple systems, dislocations on different systems interact. This results in the formation of specific distributions of dislocations. Such a distribution is characteristic of the crystal structure of the material being deformed, its stacking fault energy, the temperature of deformation, the strain and the strain rate. Precipitate and grain boundary structures also affect the distribution. In a metal of high stacking fault energy and easy cross slip, a cell structure is developed early in straining. A cell is characterized by dense tangles of dislocations arranged in walls surrounding regions or cells of low dislocation density. This can be clearly seen in Figure 9. The cell size was determined by averaging the longer and shorter dimensions parallel to the cell walls. This was found to be about one micron. The cell structure observed in titanium seems to be much more elongated than the cell structure observed in aluminum (37). Similar observation was made in an earlier study (30).

In Figure 10, we find that the substructure consists of dislocations in tangles at cell walls and regions resembling stripes running across the micrograph. By analysis of other micrographs of similar
Figure 9. Cell structure in alpha titanium deformed to 10% true strain at room temperature. (Foil normal close to [1459])
Figure 10. Dislocation structure in alpha titanium deformed to 10% true strain at room temperature. (Foil normal close to [1104])
nature of the as prestrained material, the planes of these cell walls, where there is considerable dislocation tangling, have been identified to be of the \([10\overline{1}0]\) type. The three prominent directions along which the dislocations lie were identified to be of the \(<11\overline{2}0>\) type. The projections of the three \(<11\overline{2}0>\) directions are shown by the arrows on the micrograph. In an earlier study (30), however, tangles and cell walls were frequently observed to form on planes tentatively identified as twin planes of the \([10\overline{1}2]\) type.

Mechanical twinning is sufficiently distinct from ordinary slip as to be regarded as a separate mode of deformation. The important role of twinning in plastic deformation comes from the fact that orientation changes resulting from twinning may place new slip systems in a favorable orientation with respect to the stress axis, so that additional slip can take place. Thus, twinning is important in the over-all deformation of metals with a low number of slip systems such as HCP metals. In titanium, mechanical twinning can occur on \([10\overline{1}2]\) and \([11\overline{2}n]\) \((n = 1, 2, 4)\) planes. By trace analysis, the twin seen in Figure 11 has been identified to be of the \([10\overline{1}2]\) type in agreement with the tentative identification of the twinning planes in the previous study (30). \([11\overline{2}2]\) twinning was also observed in the present study.

D. Fractional Flow Stress Recovery

The fractional flow stress recovery versus time curves in Figure 5 for the two recovery temperatures show two distinct features. An
Figure 11. Mechanical twin of the \(\{10\bar{1}2\}\) type in alpha titanium deformed to 10\% true strain at room temperature. (Foil normal close to \([\bar{2}113]\))
initial rapid increase in $f_r$ followed by leveling off. An attempt is made to correlate the observed substructure with the $f_r$ curves by considering the recovery mechanisms operating at the two temperatures separately.

1. **Fractional Flow Stress Recovery at 500°C.** The introduction of dislocations produces a large increase in the strain energy of the crystal (stored energy). This energy will be released during recovery if the dislocations either annihilate each other by mutual interactions or rearrange themselves into lower energy configurations. A considerable amount of energy may be released by the local rearrangement of the dislocations in the tangles. As seen in Figure 12 which shows recovery for 0.375 hours at 500°C, the rearrangement is not yet complete, and the cell boundaries have sharpened significantly. The observed pattern of alternating contrast strongly suggests that groups of cells deform in a cooperative way according to the constraints offered by the surroundings. In this case, at least eight alternate cells have rotated similarly during the deformation process so that they present the same crystallo-graphic axis to the electron beam. It has been found that the cell boundaries are the prism planes of the type {1010}. Comparing Figure 10 and Figure 12, we find some similarities and also some differences. The cell boundaries are more perfect in Figure 12 and the dislocations which were tangled in such positions as A in Figure 10 are absent. This suggests that upon recovering alpha-titanium for 0.375 hours at 500°C
Figure 12. Substructure in alpha titanium deformed to 10% true strain and recovered for 0.375 hours at 500°C. (Foil normal close to [1103])
annihilation of dislocation dipoles may have taken place in addition to
glide and cross slip of dislocations from such tangles. The value of
\( f_r \) corresponding to this recovery treatment is about 0.3.

As reported earlier, \{1122\} twins were also observed in the present
study. This can be seen in Figure 13. Note the presence of dislocation
loops as indicated by the arrows. The density of dislocation loops
found in the previous study (30) was considerably higher than that found
in the present study. This is probably because the temperature of
recovery is relatively high in this investigation. Clearly, there is a
tendency for some network formation such as at A. The dark line observed
to the right of the twin, marked as B, is possibly a grain boundary.

Elongated subgrains as seen in Figure 12 appear to be characteristic
of titanium recovered for 0.375 hours at 500°C. This can again be seen
in Figure 14. As in the other micrographs, the cell walls are the prism
planes. However, in this figure, regions of common diffraction contrast
are shorter. It is particularly interesting to note the tendency for
formation of some boundaries at an angle to the prism planes (see
arrows). This tendency is more clear in the figures to follow.

As can be seen from Figure 15, an increase in the recovery time
to 6 hours at 500°C (932°F) has continued to provide significant changes
in the dislocation structure. The observed tendency for boundaries to
move away from \{10\overline{1}0\} type planes is quite obvious in Figure 15, as
indicated at A. It is important to note that the subgrain A
Figure 13. Dislocation structure in alpha titanium deformed to 10% true strain and recovered for 0.375 hours at 500°C. (Foil normal close to [\(\bar{2}113\)])
Figure 14. Substructure in alpha titanium deformed to 10% true strain and recovered for 0.375 hours at 500°C. (Foil normal close to [3713])
Figure 15. Dislocation structure in alpha titanium deformed to 10% true strain and recovered for 6 hours at 500°C. The directions indicated are the projections of these crystallographic directions onto the plane of the micrograph. (Foil normal close to [01\bar{1}2])
Figure 16. Substructure in alpha titanium deformed to 10\% true strain and recovered for 6 hours at 500\°C. The directions indicated are the projections of these crystallographic directions onto the plane of the micrograph. (Foil normal close to [01\bar{1}2])
represents a section through a hexagonal prismatic figure. The faces of the hexagon have been identified to be the second order prism planes, i.e. [1120] type. A similar tendency to form subgrains by migration from the first order prism planes to the second order is seen at B. The planar trace corresponds to the [1120] plane. Comparing Figures 15 and 16, it can be seen that the dislocations are nearly straight along the directions marked on the micrograph. These directions are the projections of the [\bar{1}100] and [10\bar{1}0] crystallographic directions onto the plane of the figure. The observed dislocation structure of Figure 15 suggests the possibility of subgrains being surrounded by loops. In some cases, by tracing a dislocation line, three segments of a loop can be identified. The dislocation segments lying in the [11\bar{2}0] planes have been found to lie along the [0001] direction.

Note in Figure 15 the persistence of the dislocation loops, as indicated by arrows, even after annealing for 6 hours. This suggests the possibility that these loops are prismatic dislocation loops. These could have been formed by the jog dragging mechanism (S).

A considerable amount of energy can be released by the local rearrangement of the dislocations when low angle boundaries are formed. This process involves both cross slip and climb and thus will occur only when there is sufficient thermal activation. Also, dislocations can considerably reduce their interaction energy and density by straightening themselves (53). Such an arrangement of dislocations should be accompanied by a pronounced softening of the metal. As can be seen from
Figure 5, the $f_r$ value corresponding to recovery for 6 hours at 500°C (932°F) is about 0.6. In Figure 16, we observe the same tendency for large loop formation as in Figure 15. The dislocations seem to lie along three prominent directions. These crystallographic directions were determined to be [1100], [1010] and [0001], in accordance with the observations made in Figure 15. The tendency to form subgrains by migration from the first order prism planes to the second order can be seen at A. Note also the presence of small prismatic dislocation loops as indicated by arrows.

If what has been observed after 6 hours recovery treatment is characteristic of alpha-titanium, then an increased tendency to form hexagonal shaped subgrains should be observed for higher recovery temperatures and longer recovery times.

After 100 hours of recovery at 500°C (932°F) the subgrains are sharp and clear as can be seen in Figure 17. It is interesting to note the distinct crystallographic preference of the subboundaries. These planes on which the dislocations have arranged into low energy configurations have consistently been identified to be of the {1120} type. It is believed that the elongated collection of subboundaries indicated by the arrows in Figure 17 represent the general trace of the prism plane of the type {1010} on which the original cell structure formed (see Figure 12). Network formation was also observed in samples recovered for 100 hours at 500°C as seen at A in Figure 18. In Figure 19
Figure 17. Substructure in alpha titanium deformed to 10\% true strain and recovered for 100 hours at 500\degree C. (Foil normal close to [5506])
Figure 18. Substructure in alpha titanium deformed to 10% true strain and recovered for 100 hours at 500°C. (Foil normal close to [5506])
we can clearly observe the tendency for the subboundaries to form the hexagonal shaped subgrains as seen at B. Note the absence of small dislocation loops after 100 hours recovery.

The only change that has been observed between 6 hours and 100 hours recovery treatment at 500°C is the slight rearrangement of dislocations on the [11\overline{2}0] planes. This process probably requires cross slip and climb with an accompanying decrease in the stored energy of the metal. As seen from Figure 5, the change in $f_r$ with increasing recovery time is rather small.

The subgrain size after recovery at 500°C (932°F) for 100 hours was determined by averaging the measurements of the longer and shorter dimensions of the subgrain. This was found to be about 2.9 microns; the misorientation across the subboundaries being of the order of about 20 minutes of arc as determined from split kikuchi patterns.

2. Fractional Flow Stress Recovery at 550°C. By increasing the recovery temperature from 500°C (932°F) to 550°C (1022°F), we can expect the dislocations to rearrange into low energy stable configurations much more easily. Since the processes of cross slip and dislocation climb are both enhanced by an increase in temperature, we should be able to observe a considerable difference in structure between specimens recovered for the same length of time at 500 and 550°C. This can clearly be seen by comparing Figures 15 and 16 with Figures 20 and 21. Note the absence
Figure 19. Dislocation structure in alpha titanium deformed to 10% true strain and recovered for 100 hours at 500°C. (Foil normal close to (2111))
of dislocation loops and the sharp subgrains in Figures 20 and 21 in contrast to long and nearly straight dislocation segments observed in Figures 15 and 16.

Annihilation of prismatic dislocation loops and rearrangement of dislocations from tangles can considerably lower long range stresses. This process can account for a significant decrease in flow stress. The $f_r$ value corresponding to the recovery treatment of 6 hours duration at 550°C is found to be 0.65 (see Figure 5).

The subboundaries observed in Figure 20 are zig-zag in shape and the planes have again been identified to be of the $\{11\bar{2}0\}$ type. Apparent pairing of dislocations can be seen at A; these are probably dipoles. No explanation has yet been found for these apparent dipoles. Dislocation networks joining a subboundary on the $\{11\bar{2}0\}$ plane can be seen at A in Figure 21. By careful analysis of specimens recovered for 6 hours at 550°C, it was found that these networks formed on the basal plane. In the previous study (30), however, the predominant plane for network formation had been tentatively identified to be of the $\{10\bar{1}2\}$ type. The present observation of network formation suggests the possibility of correlating the structures observed after recovery for 6 hours at 500 and 550°C. As reported earlier, some of the long and nearly straight dislocation segments were found to lie along the $\langle 10\bar{1}0 \rangle$ type directions (see Figures 15 and 16). It is to be noted that these directions lie on the basal plane. Therefore, it is likely that the
Figure 20. Substructure in alpha titanium deformed to 10% true strain and recovered for 6 hours at 550°C. (Foil normal close to [0112])
Figure 21. Substructure in alpha titanium deformed to 10% true strain and recovered for 6 hours at 550°C. (Foil normal close to [110])
dislocation segments lying along the <10\bar10> type directions react to form dislocation networks on the basal plane. In addition, the dislocation arrays lying on the \{1\bar120\} type planes have consistently been observed to align along the [0001] direction. In the previous study (30), dislocation arrays were concluded to have formed on planes tentatively identified as deformation twin boundaries during recovery.

The large dislocation loops which surround the subgrains during recovery can now be pictured to have two of their segments lying on the basal plane and the other two segments on the \{1\bar120\} type planes. The segments originally lying along the <10\bar10> type direction react to form dislocation networks on the basal plane and the segments lying on the \{1\bar120\} type planes remain aligned along the [0001] direction. As a result of these rearrangements of dislocations from tangles, many of the subgrains formed during recovery may be pictured as bounded by twist networks on the (0001) planes and tilt boundaries on the \{1\bar120\} type planes, the total misorientation of the subgrain thus being represented by a pure twist about the [0001] direction.

Interweaving of two lower density dislocation networks to form a higher density network can be seen in Figure 22. This is a type of subgrain growth and has been called subboundary coalescence (37). The zigzag shaped subboundaries observed in Figure 20 can be seen in Figure 23 also as indicated by arrows. As before, these subboundaries were found to be of the \{1\bar120\} type.
Figure 22. Dislocation networks in α titanium deformed to 10% true strain and recovered for 6 hours at 550°C. (Foil normal close to [1123])
Figure 23. Substructure in alpha titanium deformed to 10% true strain and recovered for 100 hours at 550°C. (Foil normal close to [110])
After recovery for 100 hours at 550°C (1022°F), we can clearly see the sharpening of subgrains to form hexagonal shaped prisms in Figure 24. The sides of these subgrains have been identified to be of the [11\overline{2}0] type. Dislocations lying on the [11\overline{2}0] type planes have been found to be aligned along the [0001] direction. This is in accordance with the other observations made so far (see Figure 15). Higher magnification of a small region in Figure 24 is shown in Figure 25. The hexagonal shaped subgrain in the center of the micrograph seems to be almost separated from other subboundaries. If this subgrain persists with increasing recovery temperature, it might act as a nucleus for recrystallization.

The size of the subgrains after recovery at 550°C (1022°F) for 100 hours was determined by averaging the measurements of the longer and shorter dimensions of the subgrain. This was found to be about 3.2 microns; the misorientation across the subgrains being of the order of 30 minutes of arc. As reported earlier, the average size of the subgrains in specimens recovered at 500°C (932°F) for 100 hours was about 2.9 microns. An analysis made by Lytton (54) using the data of Young (37) suggests that the strength contributed by a subgrain structure in aluminum is inversely proportional to the subboundary spacing. If this concept is applied to the present results, the prediction is obtained that the strength due to dislocation substructure after 100 hours recovery at 500°C will be 1.10 times that for 550°C recovery. The actual ratio obtained was 1.10, providing reasonable agreement with the
Figure 24. Substructure in alpha titanium deformed to 10% true strain and recovered for 100 hours at 550°C. (Foil normal close to [1101])
Figure 25. Substructure in alpha titanium deformed to 10% true strain and recovered for 100 hours at 550°C. (Foil normal close to [110])
results of Young. Thus, the observed increase in the subgrain size can readily account for the increase in the value of the fractional flow stress recovery parameter observed for recovery at 550°C (1022°F) for 100 hours as compared to 500°C (932°F).

E. A Model of the Typical Subgrain

It is fairly clear from the present study that the subgrains formed during recovery of alpha-titanium at 500 and 550°C show a tendency to form a hexagonal prismatic shape at the temperatures investigated.

From the observations made so far, it is possible to propose a model for a typical subgrain.

It was observed that in the as-prestrained material, the cell boundaries were the prism planes of the type \{10\bar{1}0\}. In addition, the dislocations which were not tangled in the cell boundaries were found to lie along the \langle1\bar{1}20\rangle type directions. A schematic diagram of the structure of the as-prestrained material is shown in Figure 26a. The directions indicated in the figure are the projections of these crystallographic directions onto the plane of the paper. It is to be noted that the cells indicated in Figure 26a are slightly elongated rather than equiaxed.

In samples recovered for 0.375 hours at 500°C, a slight tendency for migration of cell walls from \{10\bar{1}0\} planes was observed (see arrows in Figure 14). This tendency for migration from \{10\bar{1}0\} planes to \{11\bar{2}0\} planes was more clear in samples recovered for 6 hours at 500°C.
(see at B in Figure 15 and at A in Figure 16). The first indication of the subgrains to form hexagonal shaped prisms was observed at this stage, as can be seen at A in Figure 15. Long and nearly straight dislocation segments were found to lie along the $<10\bar{1}0>$ type directions. The dislocations lying on the $\{1\bar{1}20\}$ type planes were found to align along the $[0001]$ direction. The observed dislocation structure suggested the possibility of subgrains being surrounded by loops. A schematic diagram of the formation of dislocation loops which match the experimental observations is shown in Figure 26b. The dislocation loop DEFG in Figure 26b has segments lying on the $\{1\bar{1}20\}$ and $(0001)$ planes. From the observed tendency for dislocations to form tilt boundaries on the $\{1\bar{1}20\}$ planes, the segments DE and GF represent the edge components associated with the tilt boundaries on the $\{1\bar{1}20\}$ faces of the subgrains. The segments EF and DG lie along $<10\bar{1}0>$ directions prior to reaction with loop segments from other $\{1\bar{1}20\}$ faces. After reaction in the $(0001)$ plane EF and DG would be serrated into two $<1\bar{1}20>$ directions (see A of Figure 21).

Dislocation network formation on the basal plane was observed in samples recovered for 6 hours at $550^\circ$C (see Figure 21). These networks joining a subboundary on the $\{1\bar{1}20\}$ plane (see A in Figure 21) is shown schematically at A in Figure 26c. In addition, the hexagonal shaped prisms formed after recovery for 100 hours at $550^\circ$C (see Figures 24 and 25) can also be seen. As a result of the rearrangements of dislocations during recovery many of the subgrains formed have twist networks on the
Figure 26. Schematic diagram illustrating the formation of subgrains in alpha titanium. (a) Cell structure observed after deforming to 10% true strain. (b) Formation of dislocation loops DEFG, the segments of which lie on \{1120\} and (0001) planes during recovery at 500 and 550°C. (c) Subgrains formed after recovery at 550°C for 100 hours.
(0001) planes and tilt boundaries on the {11\(\bar{2}0\}) planes as indicated in Figure 26c. The total misorientation of the subgrain can thus be represented by pure twist about the [0001] and an associated tilt \(\theta\) for the {11\(\bar{2}0\}) plane normals as shown in Figure 26c.

F. Activation Energy for Recovery

Activation energy for recovery in alpha-titanium was found to be 52.5 k cals/mole, a value fairly close to the one reported in the earlier study (30). Several authors (1,4,39,40,41) have shown that the activation energy for recovery approaches, but is generally less than that for self diffusion. Sherby and Simnad (55) have shown that the activation energy for self diffusion in metallic systems is predictable by the formula \(Q = RT_m (K_0 + V)\) where \(Q\) is the activation energy for self diffusion, \(R\) is the gas constant, \(T_m\) is the absolute melting temperature, \(K_0\) is the crystal structure factor equal to 17 for FCC and HCP metals, and \(V\) is the valence. Using a valence of four for titanium, the predicted activation energy for self diffusion is 87.0 k cals/mole. The observed activation energy for recovery is about 60% of that for self diffusion. This proportion is in fair agreement with other authors (1,4,39,40,41).
V. CONCLUSIONS

In an attempt to correlate the dislocation structure with the plastic deformation and flow stress recovery of commercial purity polycrystalline alpha-titanium, the following conclusions were reached:

1. After straining the metal to 10% true strain, well developed cell structure results. These cells were slightly elongated rather than equiaxed. The cell boundaries were identified to be \{10\overline{1}0\} type planes. Nearly straight dislocation segments formed along \langle11\overline{2}0\rangle directions.

2. Fractional flow stress recovery of alpha-titanium proceeded in two stages, an initial rapid recovery followed by a leveling off. This was true for recovery at both 500°C (932°F) and 550°C (1022°F).

3. Observation of extensive network formation without stacking faults suggests that alpha-titanium has a rather high stacking fault energy.

4. The subgrains appeared to be made up of dislocation loops. As a result of the rearrangements and interactions of dislocations during recovery, many of the subgrains formed were bounded by twist networks on the (0001) planes and tilt boundaries on the \{11\overline{2}0\} type planes, the total misorientation of the subgrains thus being represented by a pure twist about the [0001] direction. There was a marked tendency to form hexagonal shaped subgrains.
5. The activation energy for the flow stress recovery was found to be 52.5 k cals/mole. This value is about 60% of the predicted value of activation energy for self diffusion.
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TABLE I

Formulae Useful in HCP Crystallography
(From axis-four index notation)

<p>| | | | |</p>
<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>1. Indices of the direction ([\text{defg}]) normal to plane ((\text{hkil}))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>([d,e,f,g] = (h,k,l,(3/2)(a/c)^2))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2. Criteria for direction ([\text{defg}]) to lie in plane ((\text{hkil}))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Also can be used to solve for line or intersection of planes</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(2dh + dk + 2ek + eh + gl = 0)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>3. Angle between directions ((\text{hkil})) and ([\text{defg}])</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(\cos \theta = \frac{hd + ke + (1/2)(kd + he) + (1/3)gl}{\sqrt{\left(d^2 + e^2 + de + (1/3)g^2 \left(\frac{a}{c}\right)^2 \left(h^2 + k^2 +hk + (1/3)1^2 \left(\frac{a}{c}\right)^2\right)\right)^{1/2}}} )</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>4. Angle between planes ((\text{hkil})) and ([\text{defg}])</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(\cos \theta = \frac{hd + ke + (1/2)(kd + he) + (3/4)gl}{\sqrt{\left(d^2 + e^2 + de + (3/4)g^2 \left(\frac{a}{c}\right)^2 \left(h^2 + k^2 +hk + (3/4)1^2 \left(\frac{a}{c}\right)^2\right)\right)^{1/2}}} )</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>5. Length OP of ([\text{hkil}])</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(OP = \sqrt{3a^2(h^2 + k^2 +hk) + c^2l^2}^{1/2})</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>6. Three axis direction ([UWV]) to four axis direction ([UV\text{t}W])</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(U = u-t, \ V = v-t, \ W = w)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>or (u = (1/3)(2U-V), \ v = (1/3)(2V-U), \ w = W, \ t = -(u+v))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(Four-axis-three index notation)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>7. Zone axis of any two planes ((h_1k_1l_1)) and ((h_2k_2l_2)) is given by</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>([u v w] = 1_2(2k_1 + h_1)<em>{-1_1}(2k_2 + h_2); 1_2(2h + k_1)</em>{-1_1}(2h + k_1))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(3(h_1k_2 - h_2k_1))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Using (1), plane perpendicular to ([u v w]) for alpha titanium has the indices ((u v \cdot (5/3)w))</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
TABLE II

Data for Calculation of $f_r$ at 500°C (932°F)

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>$\sigma_y$ (ksi)</th>
<th>$\sigma_1$ (ksi)</th>
<th>$\sigma_2$ (ksi)</th>
<th>Recovery Time (hours)</th>
<th>$f_r$</th>
</tr>
</thead>
<tbody>
<tr>
<td>40A, 16B2</td>
<td>22.23</td>
<td>43.03</td>
<td>36.85</td>
<td>0.375</td>
<td>0.297</td>
</tr>
<tr>
<td>40A, 9B7</td>
<td>21.78</td>
<td>42.64</td>
<td>35.12</td>
<td>0.75</td>
<td>0.360</td>
</tr>
<tr>
<td>40A, 16B4</td>
<td>21.54</td>
<td>41.85</td>
<td>31.66</td>
<td>1.5</td>
<td>0.501</td>
</tr>
<tr>
<td>19A, 2B4</td>
<td>22.26</td>
<td>40.63</td>
<td>30.69</td>
<td>3.0</td>
<td>0.541</td>
</tr>
<tr>
<td>29A, 3B4</td>
<td>23.38</td>
<td>42.70</td>
<td>31.13</td>
<td>6.0</td>
<td>0.599</td>
</tr>
<tr>
<td>38A, 15B1</td>
<td>23.01</td>
<td>43.25</td>
<td>32.05</td>
<td>12.0</td>
<td>0.553</td>
</tr>
<tr>
<td>27A, 7B7</td>
<td>24.78</td>
<td>43.14</td>
<td>31.02</td>
<td>12.0</td>
<td>0.660</td>
</tr>
<tr>
<td>33A, 15B4</td>
<td>22.45</td>
<td>43.21</td>
<td>31.57</td>
<td>12.0</td>
<td>0.560</td>
</tr>
<tr>
<td>25A, 5B5</td>
<td>22.18</td>
<td>43.24</td>
<td>29.97</td>
<td>24.0</td>
<td>0.629</td>
</tr>
<tr>
<td>29A, 6B5</td>
<td>22.90</td>
<td>42.10</td>
<td>29.73</td>
<td>48.0</td>
<td>0.644</td>
</tr>
<tr>
<td>29A, 8B4</td>
<td>22.50</td>
<td>42.24</td>
<td>28.56</td>
<td>100.0</td>
<td>0.692</td>
</tr>
<tr>
<td>27A, 8B7</td>
<td>22.34</td>
<td>41.07</td>
<td>28.18</td>
<td>100.0</td>
<td>0.688</td>
</tr>
</tbody>
</table>
TABLE III

Data for Calculation of $f_r$ at 550$^\circ$C (1022$^\circ$F)

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>$\sigma_y$ (ksi)</th>
<th>$\sigma_1$ (ksi)</th>
<th>$\sigma_2$ (ksi)</th>
<th>Recovery Time(hours)</th>
<th>$f_r$</th>
</tr>
</thead>
<tbody>
<tr>
<td>41A, 16811</td>
<td>21.89</td>
<td>42.37</td>
<td>32.99</td>
<td>0.375</td>
<td>0.458</td>
</tr>
<tr>
<td>40A, 1689</td>
<td>21.53</td>
<td>42.44</td>
<td>31.50</td>
<td>0.75</td>
<td>0.523</td>
</tr>
<tr>
<td>40A, 1687</td>
<td>22.10</td>
<td>42.32</td>
<td>30.63</td>
<td>1.5</td>
<td>0.578</td>
</tr>
<tr>
<td>34A, 1081</td>
<td>25.51</td>
<td>44.61</td>
<td>32.54</td>
<td>3.0</td>
<td>0.631</td>
</tr>
<tr>
<td>37A, 1183</td>
<td>21.37</td>
<td>42.29</td>
<td>28.94</td>
<td>6.0</td>
<td>0.638</td>
</tr>
<tr>
<td>37A, 1186</td>
<td>21.44</td>
<td>42.48</td>
<td>28.98</td>
<td>6.0</td>
<td>0.641</td>
</tr>
<tr>
<td>42A, 1281</td>
<td>21.34</td>
<td>42.49</td>
<td>29.47</td>
<td>12.0</td>
<td>0.615</td>
</tr>
<tr>
<td>42A, 1282</td>
<td>21.82</td>
<td>41.95</td>
<td>28.37</td>
<td>12.0</td>
<td>0.674</td>
</tr>
<tr>
<td>35A, 1284</td>
<td>22.80</td>
<td>42.57</td>
<td>28.73</td>
<td>12.0</td>
<td>0.700</td>
</tr>
<tr>
<td>41A, 1286</td>
<td>21.80</td>
<td>42.29</td>
<td>27.87</td>
<td>24.0</td>
<td>0.703</td>
</tr>
<tr>
<td>42A, 1287</td>
<td>21.95</td>
<td>42.79</td>
<td>28.15</td>
<td>24.0</td>
<td>0.700</td>
</tr>
<tr>
<td>35A, 1483</td>
<td>22.22</td>
<td>43.06</td>
<td>28.75</td>
<td>48.0</td>
<td>0.686</td>
</tr>
<tr>
<td>38A, 1484</td>
<td>21.55</td>
<td>41.75</td>
<td>28.09</td>
<td>48.0</td>
<td>0.675</td>
</tr>
<tr>
<td>35A, 1486</td>
<td>23.45</td>
<td>43.77</td>
<td>29.90</td>
<td>100.0</td>
<td>0.682</td>
</tr>
<tr>
<td>35A, 1487</td>
<td>22.03</td>
<td>42.55</td>
<td>26.40</td>
<td>100.0</td>
<td>0.787</td>
</tr>
</tbody>
</table>
RELATIONSHIP BETWEEN FLOW STRESS RECOVERY AND DISLOCATION STRUCTURE
IN POLYCRYSTALLINE ALPHA-TITANIUM

by

P. Ganendra Nath Pawar

ABSTRACT

Cell structure was observed in polycrystalline alpha-titanium deformed up to 10% true strain at room temperature. These cells appeared to be slightly elongated rather than equiaxed. The cell boundaries were identified to be the \( \{10\bar{1}0\} \) type planes.

Flow stress recovery of alpha titanium at 500 and 550\(^{\circ}\)C proceeded in two stages: (1) an initial rapid recovery during the first 6 hours and (2) a leveling off after 6 hours. Dislocation structure was studied by using transmission electron microscopy.

During the first stage of recovery, significant rearrangement of dislocations took place. At this stage fairly regular dislocation loops appeared to surround the subgrains. The segments of the dislocation loop were found to lie along \(<10\bar{1}0>\) type directions on the \((0001)\) plane, and \([0001]\) type directions on the \(\{1\bar{1}20\}\) planes. In the second stage, continued rearrangement of dislocations formed low angle boundaries. The segments of dislocations lying along the \(<0\bar{1}0\bar{1}>\) type direction reacted to form networks on the basal plane, and those lying on the \(\{1\bar{1}20\}\) planes continued to remain aligned along \([0001]\) direction. It was postulated that some of the subgrains formed as a
result of rearrangement and interaction of dislocations developed
twist networks on the (0001) planes and tilt boundaries on the \{11\bar{2}0\}
type planes.

An inverse relationship between the strength contributed by sub-
grain structure and the subboundary spacing predicted an increase in
the subgrain size at the higher recovery temperature. The measured
and the predicted value were in reasonably good agreement.