

# The Effect of Hydrogen as a Temporary Alloying Element on the Microstructure and Tensile Properties of Ti-6Al-4V

WILLIAM R. KERR

Ti-6Al-4V alloy, to which 0.6 wt pct to 1.0 wt pct (22 to 33 at. pct) hydrogen has been added, can undergo a phase transformation which produces unique, fine microstructures. Specimens of the alloy were heated to 870 °C, transformed at temperatures between 540 °C and 700 °C, and the microstructures were determined as a function of hydrogen content and transformation temperature. Microstructures and tensile properties of sheet specimens were determined after such transformation followed by dehydrogenation at temperatures between 650 °C and 760 °C. The highest yield strength (1130 MPa) and good ductility (9 pct El) were associated with a fine equiaxed microstructure obtained in material charged with approximately 1.0 wt pct hydrogen, transformed at 565 °C and dehydrogenated at 675 °C. Lower strengths and ductilities were associated with acicular microstructures produced by transformation at higher temperatures or coarser structures produced at higher dehydrogenation temperatures.

## I. INTRODUCTION

TITANIUM and titanium alloys have a large affinity for hydrogen. Pure titanium will absorb more than 60 at. pct (3 wt pct) hydrogen at one atmosphere pressure and 640 °C. Still greater concentrations are absorbed at lower temperatures.<sup>1,2</sup> The most obvious and most studied effect of hydrogen on titanium alloys is embrittlement at ambient temperature.<sup>3,4,5</sup> Unlike other gases, the absorption of hydrogen is reversible. The diffusivity of hydrogen is sufficiently high,<sup>6</sup> and its equilibrium concentration at easily achieved vacuum pressures is sufficiently low, so that hydrogen can be removed by vacuum annealing.

The ease with which hydrogen can be added to and removed from titanium has led to the utilization of hydrogen as a temporary alloying element. Room temperature embrittlement as a result of sufficiently high hydrogen contents provides an economical method for producing titanium powder by attrition.<sup>7</sup> Hydrogen is removed from the powder and ductility is restored by vacuum annealing. Several studies have shown improvements in the hot workability of titanium alloys charged with high concentrations of hydrogen.<sup>7-12</sup> Increases in plasticity were attributed to enhanced self-diffusion of titanium in the presence of hydrogen, and decreases in forging loads were related with hydrogen stabilization of the more hot-workable bcc beta phase.

Hydrogen in titanium alloys may also be used to effect a phase transformation which results in unique, fine microstructures. Above 815 °C, Ti-6Al-4V containing 0.4 to 1.35 wt pct hydrogen transforms completely to beta. When specimens of this alloy containing 0.46, 0.93, and 1.35 wt pct hydrogen were water quenched from the beta phase field (870 °C), the phases identified were, respectively: primary alpha plus orthorhombic martensite ( $\alpha + a''_H$ ); orthorhombic martensite plus beta saturated with hydrogen ( $a''_H + B_H$ ); and beta saturated with hydrogen ( $B_H$ ). No hydrides were

detected either by optical microscopy or by X-ray diffraction. When heated to the beta phase field then step cooled to 650 °C and held for various times, the 0.93 wt pct hydrogen specimen underwent a transformation which resulted in a finely divided mixture of three phases identified as  $\alpha$ ,  $\beta$ , and hydride.<sup>13</sup> The transformation suggests a three phase eutectoid in which  $\beta_H \rightarrow \alpha + \text{hydride}$ .

This reaction has been observed in Ti-6Al-4V containing 0.53 to 1.35 wt pct hydrogen and for isothermal transformation temperature in the range 540 °C to 700 °C. Dehydrogenation, at appropriate temperatures, of the finely divided eutectoid transformation structure results in unique microstructures consisting of essentially equiaxed particles of alpha and beta. Transformation kinetics as determined by optical microscopy were reported in Reference 12. However, the very fine microstructures were difficult to resolve optically. In this study, the microstructures of fully transformed material and of fully transformed and dehydrogenated material were determined as a function of hydrogen content and of transformation and dehydrogenation temperatures, using scanning electron microscopy. Sheet tensile specimens charged with hydrogen, then heat treated and dehydrogenated to produce selected microstructures were used to determine mechanical property, microstructure, and heat treatment correlations.

## II. EXPERIMENTAL PROCEDURE

### A. Material

The starting material used in this investigation was mill annealed titanium 6Al-4V sheet, 3.3 mm (0.130 inch) thick, having the analysis shown in Table I. Both metallographic specimens and tensile blanks were sheared from this sheet. Tensile specimens were machined from the blanks before hydrogenation with the tensile axis oriented parallel to the rolling direction of the sheet.

### B. Hydrogenation

Specimens were hydrogenated in a stainless steel tube, 75 mm (3 inches) in diameter by 1.5 m (5 feet) long, sealed

WILLIAM R. KERR is Metallurgist with High Temperature Materials Branch, Metals and Ceramics Division, Air Force Wright Aeronautical Laboratories/Materials Laboratory, Wright-Patterson Air Force Base, OH 45433.

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**Table I. Ingot Chemistry of Material Used in This Investigation, Pct by Weight**

	Al	V	Fe	N	O	H
Top	6.4	4.4	0.18	0.013	0.14	0.015
Bottom	6.5	4.2	0.13	0.012	0.13	0.022

by flanges at both ends and mounted in a laboratory furnace. Specimens to be hydrogenated were placed in a loose-fitting box made of Ti-6Al-4V sheet, 0.42 mm (0.016 inch) thick. A number of trial runs indicated that this box reduced contamination of the specimens from residual impurities in the hydrogenating gas to a negligible amount and aided in uniformity of hydrogen absorption during the run. The box containing the specimens was placed in the work zone at the center of the furnace where the temperature was uniform within  $\pm 10^\circ\text{C}$  over a length of 250 mm (10 inches).

The furnace was sealed and subjected to three purging cycles by pumping down to approximately 10 torr and back-filling with argon. The furnace was then heated to the hydrogenation temperature,  $650^\circ\text{C}$ , with an atmosphere of flowing argon at a slight positive pressure in the tube. A flow of hydrogen was then added to the argon stream and the partial pressures of hydrogen and argon adjusted by means of Linde Flowmeters accurate to within  $\pm 10$  pct. A slight positive pressure was maintained in the hydrogenation system at all times to prevent air entering the system. The effluent gas was passed through a vacuum oil bubbler for the same reason. Excess hydrogen was burned off by passing the effluent gas through a Bunsen burner.

The hydrogen content of the specimens was controlled by the time of hydrogenation, the partial pressure of hydrogen in the hydrogenation gas, and an efficiency factor, defined as the ratio of the amount of hydrogen retained by the specimens to the amount which flowed through the chamber during the run. Values of partial pressure, time, and efficiency were determined empirically. Hydrogen contents varied by  $\pm 5$  pct within specimens and between specimens within the same run, and by  $\pm 15$  pct between nominally similar runs. No variations in microstructure which could be attributed to variations in hydrogen content were observed within any individual specimen.

At the end of the predetermined hydrogenation time, a third gas, Ar + 4 pct H, which is nonflammable, but which represents a partial pressure of hydrogen in equilibrium with a hydrogen content of approximately 0.6 wt pct in Ti-6Al-4V at  $650^\circ\text{C}$ , was added to the gas stream. Hydrogen and argon were then shut off and the outlet from the furnace was closed. A pressure of 1.1 to 1.2 atmospheres was maintained in the system, and specimens were allowed to equilibrate for one hour at temperature and then cooled overnight. While some hydrogen transfer between the specimens and the gas phase undoubtedly took place during equilibration and cooling, preliminary studies showed the gain or loss was not significant.

The hydrogen contents of specimens were determined by weighing specimens before and after hydrogenation to the nearest 0.0001 g. A check of this technique on selected tensile specimens by vacuum fusion analysis and by weight change before and after vacuum annealing indicated that it

was accurate within  $\pm 2$  pct of the hydrogen present in the specimens. Specimens which were weighed, hydrogenated, reweighed, heat treated, dehydrogenated, and weighed again showed slight weight increases. Vacuum fusion analyses of these specimens indicated very low (10 ppm or less) hydrogen contents. The net weight gain (always  $< 2$  pct of the hydrogen content) is probably a result of some oxygen pick-up ( $< 300$  ppm) during the cycle.

After hydrogenation and encapsulation in quartz tubes to prevent oxidation, all specimens were beta treated by heating to  $870^\circ\text{C}$ , which is above the beta transus for Ti-6Al-4V with the hydrogen contents considered here.<sup>12</sup> The specimens were held at this temperature for one-half hour. Most of the specimens were then transferred to furnaces at lower temperatures for transformation, then air cooled to room temperature while still sealed in the quartz tubes. A few specimens were cooled to room temperature after the  $870^\circ\text{C}$  treatment, reheated to the transformation temperature for predetermined times, and air cooled.

### C. Microscopy

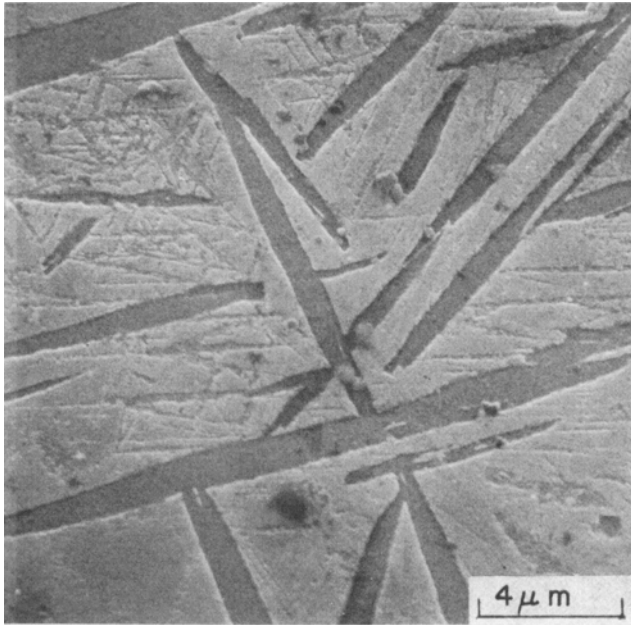
All metallographic specimens were ground and polished by conventional techniques and etched with Kroll's etch. An earlier study<sup>12</sup> using optical microscopy determined the effect of time, temperature, and hydrogen content on the extent of the transformation which occurs in Ti-6Al-4V-H. In the present study, only heat treatment times resulting in complete transformation were considered. Because of difficulty in resolving the resulting microstructures by optical microscopy, scanning electron microscopy was used in this study. Imaging was achieved with back-scattered electrons. This reverses the usual shading observed in optical microscopy of titanium alloys; alpha, which is low in vanadium and high in aluminum, has a lower average atomic number than beta and therefore appears darker. For a few specimens, substructures were determined by transmission electron microscopy.

### D. Dehydrogenation

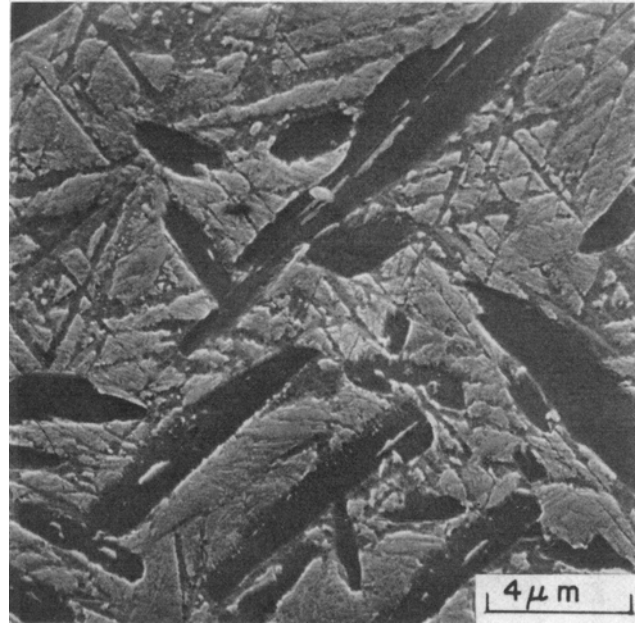
After heat treatment, the tensile specimens were removed from the quartz tubes and dehydrogenated in a Centorr vacuum furnace. To minimize surface contamination, specimens were suspended in a stainless steel tube which had its ends covered by loose-fitting titanium sheets. Temperature was measured by a thermocouple placed inside the tube, adjacent to one of the specimens. The furnace was pumped down to a  $5 \times 10^{-5}$  torr or less, then heated to and maintained at the dehydrogenation temperature until the chamber pressure again reached  $5 \times 10^{-5}$  torr or less.

### E. Tensile Tests

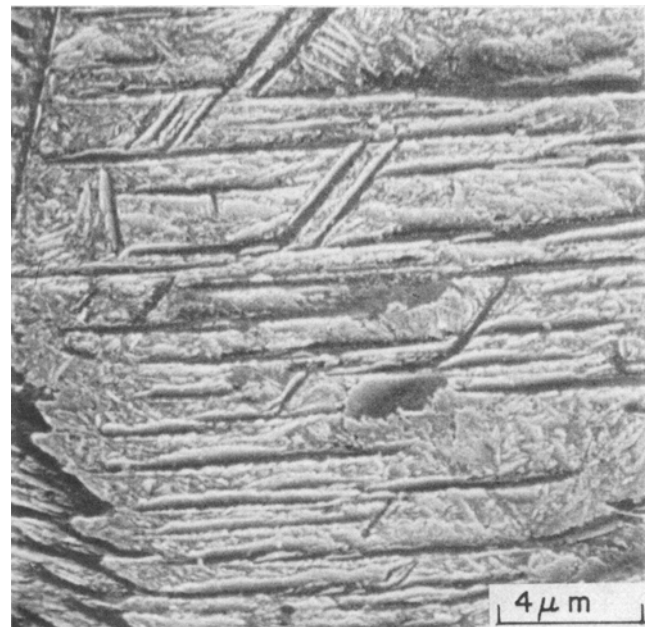
All tensile testing was done on an Instron machine using a 10,000-pound load cell and a cross-head speed of 0.002 cm per second. Tensile specimens were 15 mm (0.625 inch) by 200 mm (8 inches), having a reduced gage section 38 mm (1.5 inches) long by 9.5 mm (0.375 inch) wide by 3.3 mm (0.130 inch) thick. Duplicate tests were run for all conditions.



(a)



(b)



(c)

Fig. 1—Microstructure of Ti-6Al-4V-H alloys beta treated at 870 °C, cooled to a relatively high transformation temperature: (a) 0.45 wt pct H, transformed at 650 °C. (b) 0.53 wt pct H, transformed at 700 °C. (c) 0.93 wt pct H, transformed at 700 °C.

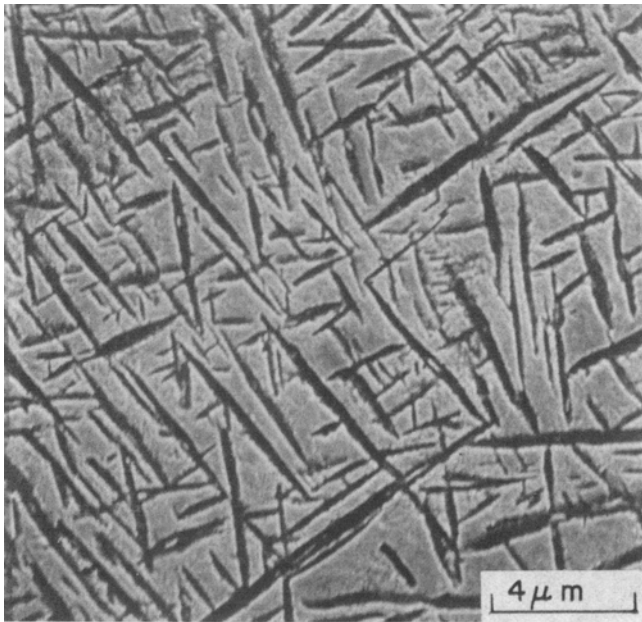
### III. RESULTS AND DISCUSSION

#### A. As-Hydrogenated and Transformed Microstructures

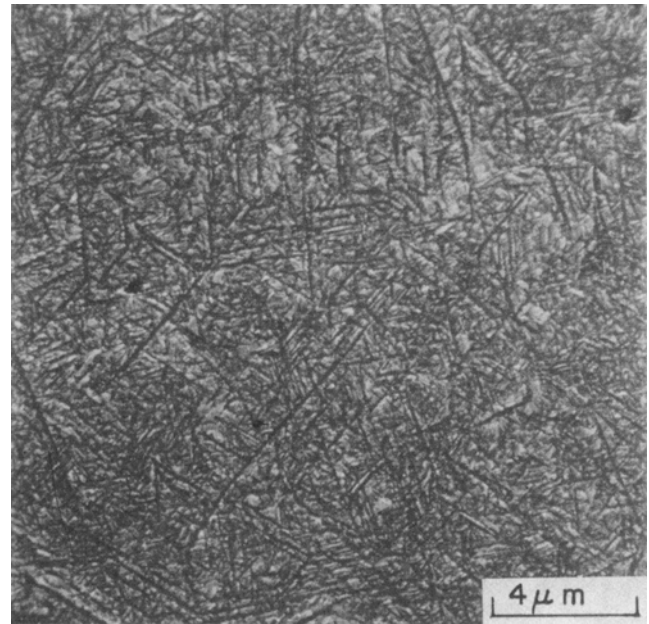
At higher transformation temperatures, relatively coarse, acicular phases result as shown in Figure 1. In the specimen containing 0.45 wt pct hydrogen (Figure 1(a)) the three phase transformation does not occur indicating that this hydrogen concentration lies outside the eutectoid phase field.<sup>12</sup> The microstructure consists of primary alpha, formed above the  $M_s$ , in a matrix which X-ray diffraction indicates is orthorhombic martensite. In the specimen containing 0.53 wt pct hydrogen, transformed at 700 °C (Figure 1(b)), the primary phase is again alpha, but X-ray diffraction indicates that the matrix is the eutectoid transformation product reported for higher hydrogen contents by Reference 13. For the same transformation temperature, a specimen containing 0.73 wt pct hydrogen evidenced a structure similar to Figure 1(b) but showed less of the primary phase. Figure 1(c) shows the microstructure of a specimen containing 0.93 wt pct hydrogen, consisting almost entirely of the transformation phases. The primary

phase in this specimen may be either alpha or hydride. Transformation at 700 °C of a specimen containing 1.35 wt pct hydrogen resulted in a microstructure similar to that of Figure 1(c) but with more of the primary phase, undoubtedly hydride since the amount increases with hydrogen content.

Much finer structures were obtained at lower transformation temperatures. The microstructures resulting from transformation at 590 °C of specimens containing 0.56, 0.73, and 1.35 wt pct hydrogen are shown in Figure 2. For the lowest hydrogen content (Figure 2(a)) the acicular primary alpha is somewhat finer than that of Figure 1(b),

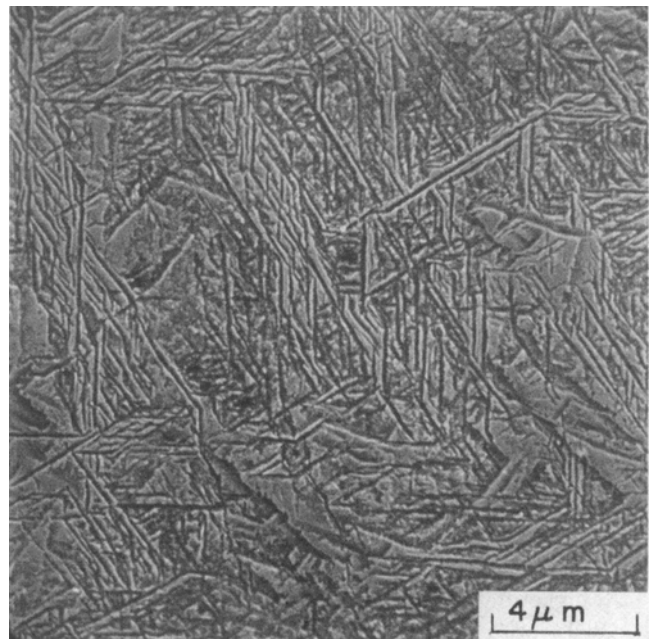


(a)



(b)

Fig. 2—Microstructure of Ti-6Al-4V-H alloys beta treated at 870 °C, cooled to 590 °C for transformation: (a) 0.56 wt pct H; (b) 0.73 wt pct H; (c) 1.35 wt pct H.

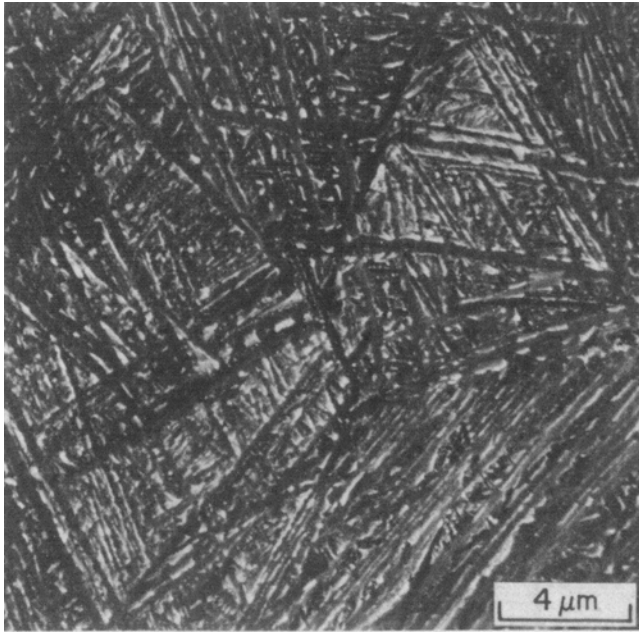


(c)

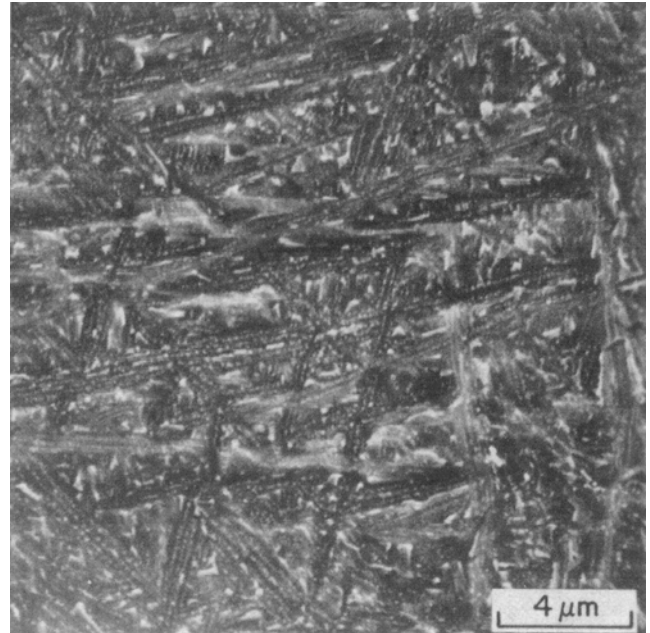
which has the same nominal hydrogen content. The specimen containing 0.73 wt pct hydrogen (Figure 2(b)) appears to consist only of the eutectoid transformation product; presumably formation of primary alpha is suppressed on cooling to this lower temperature. A microstructure essentially the same as that of Figure 2(b) was obtained in the specimen containing 0.93 wt pct hydrogen transformed at 590 °C. In a specimen containing 1.35 wt pct hydrogen (Figure 2(c)) a somewhat coarser structure resulted because of the formation of primary hydride platelets.

Encapsulated specimens containing 0.93 wt pct hydrogen which were beta treated then air cooled to room temperature showed no evidence of the transformation or of hydrides when examined by optical microscopy. Estimates based on the kinetic data of Reference 12 indicate that no transformation would have occurred during cooling. Specimens which were similarly treated then reheated to 590 °C and transformed had the same microstructure as that shown in Figure 2(b). For this procedure the transformation time was much faster, 2 hours, rather than 4 hours for cooling directly to the transformation temperature after beta treatment.

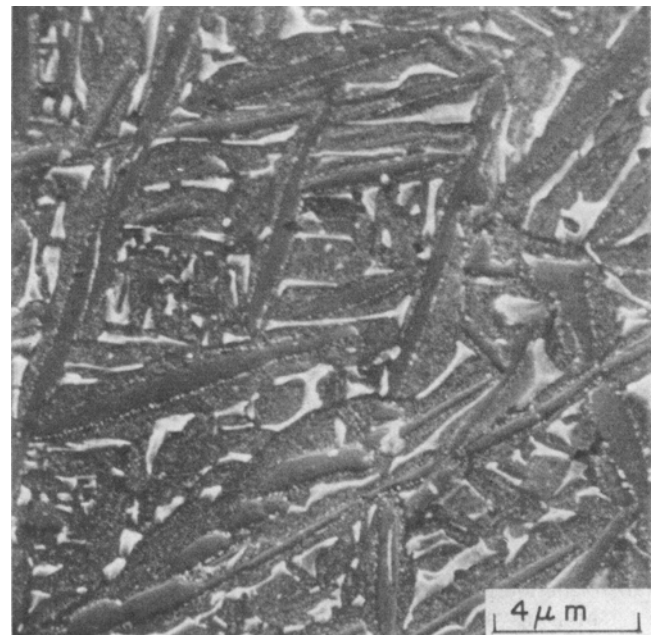
Equilibrium relationships in this nominally quaternary system were not determined, only the hydrogen contents and temperatures for which transformation occurred. The formation of primary alpha at lower hydrogen concentrations, of primary hydride at higher hydrogen concentrations and of a finely divided transformation product in the vicinity of 0.9 wt pct hydrogen is consistent with the eutectoid transformation suggested by Reference 13. These data and those of Reference 12 indicate that aluminum and vanadium shift the hydrogen eutectoid to higher temperatures than for unalloyed titanium.<sup>2</sup> Further, the three-phase eutectoid phase field is no longer isothermal and does not go to completion at the transformation temperatures investigated; *i.e.*, some



(a)



(b)



(c)

Fig. 3—Microstructure of Ti-6Al-4V-H alloys beta treated at 870 °C, subjected to various transformation and dehydrogenation treatments: (a) 0.43 wt pct H, transformed at 540 °C, dehydrogenated at 650 °C; (b) 0.62 wt pct H, transformed at 590 °C, dehydrogenated at 700 °C; (c) 1.0 wt pct H, transformed and dehydrogenated at 700 °C.

of the beta remains untransformed and is stable or at least metastable to room temperature as indicated by optical microscopy.<sup>13</sup>

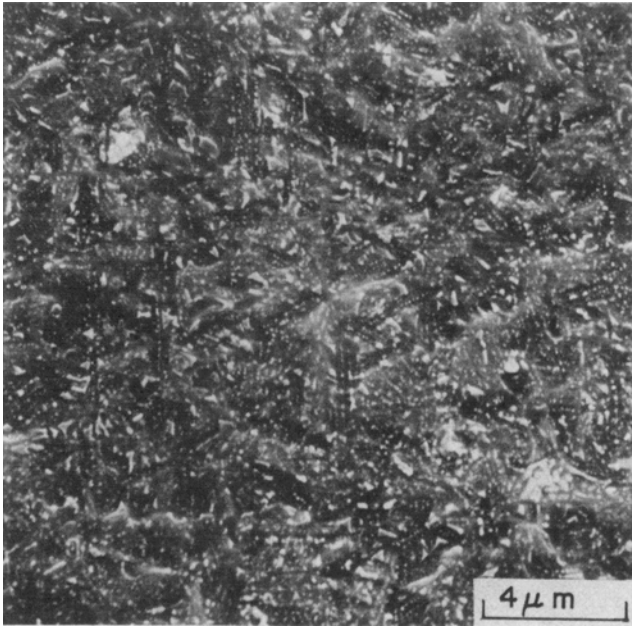
#### B. Dehydrogenated Microstructures

In addition to the variables of hydrogen content and transformation temperature, microstructure is also influenced by dehydrogenation temperature. For this study, dehydrogenation was limited to the temperature range 650 °C to 760 °C; at lower temperature hydrogen evolution was very slow and at higher temperatures coarsening of the fine structure was excessive.

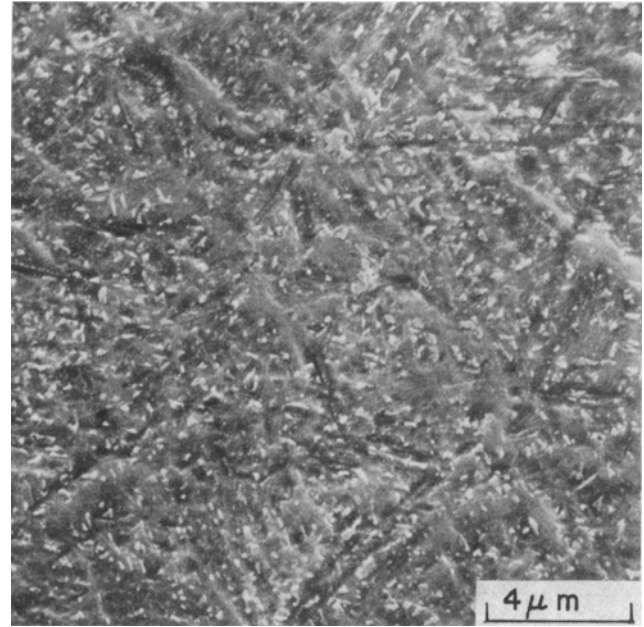
The transformation structure tends to be reflected in the dehydrogenated microstructures as shown in Figure 3. The structure of a specimen containing 0.43 wt pct hydrogen heat treated at 540 °C to produce a martensitic structure and dehydrogenated at 650 °C is shown in Figure 3(a). During dehydrogenation alpha was formed isothermally at a relatively low temperature as hydrogen was removed. The result is acicular alpha, but the particles are much finer than would

be produced by quenching a nonhydrogenated material from the beta phase. The microstructure of Figure 3(b), for a specimen containing 0.62 wt pct hydrogen transformed at 590 °C and dehydrogenated at 700 °C, shows evidence of primary alpha in a matrix of dehydrogenated transformation product and is comparable to the as-transformed structure of Figure 2(a). Similarly, Figure 3(c) showing the microstructure of a specimen containing 1.0 wt pct hydrogen both transformed and dehydrogenated at 700 °C can be compared to Figure 1(c).

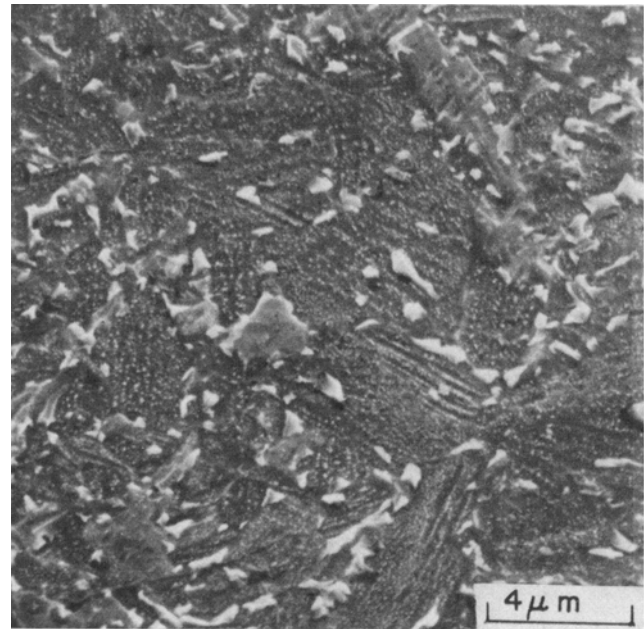
Figure 4 shows the effect of dehydrogenation temperature on specimens containing 1.0 wt pct hydrogen transformed



(a)



(b)



(c)

Fig. 4—Microstructure of Ti-6Al-4V-1H alloys beta treated at 870 °C, transformed at 590 °C, dehydrogenated at various temperatures: (a) 650 °C; (b) 700 °C; (c) 760 °C.

at 590 °C. The micrograph of Figure 4(a) for a specimen dehydrogenated at 650 °C is typical of the finest structures produced in this study. Some growth of the phases occurs at 700 °C as shown in Figure 4(b), and significant growth occurs at higher temperature, 760 °C, as shown in Figure 4(c).

The microstructure of the dehydrogenated three-phase transformation product consists of particles of light-colored beta, high in vanadium and low in aluminum, in a matrix which is primarily alpha. The variation in size and shape of these beta particles with transformation temperature (compare Figures 3(c) and 4(d)) indicates that they are formed, or at least nucleated during the transformation. The partitioning of alloying elements during transformation is sufficient so that a significant amount of beta is stable or at least metastable on cooling to room temperature. During dehydrogenation, the proportion of beta stabilized by hydrogen transforms to alpha, but that proportion stabilized by the partitioning of vanadium during transformation remains. Because neither vanadium nor aluminum forms stable hydrides at temperatures as high as those used for transformation, the hydride phase probably contains very low concentrations of these elements. During dehydrogenation the hydride transforms to alpha, the stable phase in unalloyed titanium at these temperatures. The alpha particles

already present would form nuclei for the growth of additional alpha from the hydride. Thus, by employing the eutectoid transformation and subsequent dehydrogenation, microstructures consisting of equiaxed alpha, which grow by outward diffusion of hydrogen, can be obtained. Growth of the alpha is limited by the presence of numerous small beta particles. This is in contrast to the acicular alpha which forms by Widmanstätten precipitation in a matrix of beta as in conventional heat treatment.

### C. Tensile Results

Table II shows the result of tensile tests on specimens hydrogenated to various levels and subjected to various

**Table II. Hydrogen Contents, Heat Treatments, and Tensile Properties for the Ti-6Al-4V Specimens**

H <sub>2</sub> Wt Pct	Transformation		Dehydrogenation		$\sigma_u$ MPa	$\sigma_y$ MPa	Pct El in 38 mm	R. A. Pct
	Temp. °C	Time Hours	Temp. °C	Time Hours				
0.00	590	4	700	6	990	930	12.7	32.1
0.43	540	2	650	6	1140	1040	9.6	22.5
0.46	540	2	760	5	1060	960	12.3	28.4
0.43	590	2	650	6	1140	1030	5.8	9.2
0.61	540	16	700	6	1100	1010	11.8	32.2
0.62	590	4	700	6	1100	1010	12.2	35.3
0.62	565	8	675	8	1140	1050	10.7	28.5
0.85	540	16	700	6	1140	1100	11.6	34.7
0.86	565	8	675	8	1220	1180	9.8	26.2
0.85	590	4	700	6	1150	1120	11.6	34.3
0.94	540	16	700	6	1130	1100	11.8	32.4
0.95	565	8	675	8	1220	1180	9.4	27.1
0.98	590	4	650	6	1230	1170	3.6	6.8
0.97	590	4	700	6	1150	1120	12.6	45.4
0.97	590	4	760	3.5	1080	1040	14.2	47.7
0.98	590*	2	650	6	1250	1170	4.0	7.6
0.99	590*	2	700	6	1160	1110	11.9	37.0
0.99	590*	2	760	3.5	1080	1030	14.0	43.4
0.96	700	8	650	6	940	820	9.4	12.4
0.95	590	4	675	8	1210	1170	8.6	16.4
0.96	700	8	700	6	940	830	9.8	12.6
0.97	700	8	760	3.5	940	840	10.3	13.8
1.25	590	8	650	6	1220	1180	1.8	3.5
1.25	590	8	760	3	1060	1020	14.0	47.1
1.25	590*	8	650	6	1230	1200	2.1	2.3
	As-received, mill annealed				1030	1000	11.1	29.8
	Typical annealed, Reference 14				930	900	12.4	
	Typical S.T.A., Reference 14				1220	1130	6.4	

\*Indicates specimens air cooled to room temperature between beta treatment and transformation.

transformation and dehydrogenation temperatures in order to assess the effect on mechanical properties of these three variables. For each condition duplicate specimens were tested; reproducibility between specimens within each condition was very good. For both yield and ultimate stresses the values for the two specimens within a condition varied only  $\pm 1$  pct. For both elongation and reduction in area the average variation was less than  $\pm 7$  pct of the average value. The greatest variations occurred at low ductility values where the maximum absolute difference was  $\pm 2$  pct. Within the table results are arranged in order of increasing nominal hydrogen content, then in order of increasing transformation temperature, and finally in order of increasing dehydrogenation temperature.

For comparison, tensile data for as-received material and an average value for annealed and solution treated and aged Ti-6Al-4V<sup>14</sup> have been included in Table II. The as-received strength is on the high side of the range for mill annealed material but since there is nothing unusual in the chemistry of this heat, the high strength probably indicates that the material was not completely annealed as received.<sup>15</sup> One set of specimens without hydrogen additions was subjected to the same thermal treatment as hydrogenated specimens which showed the most improvement in tensile properties in order to assess the effect of the thermal cycle alone. In the

absence of hydrogen this treatment amounts to an annealing treatment, and the resulting properties are close to average values for annealed Ti-6Al-4V.

Figure 5 shows the yield strength and elongation as a function of hydrogen content for various transformation and dehydrogenation treatments. Some increase in yield strength was obtained for the specimens processed with 0.45 and 0.62 wt pct hydrogen. For the 0.45 wt pct hydrogen condition the microstructure consisted of fine acicular alpha (Figure 3(a)). For the higher, (0.62 wt pct) hydrogen content, the microstructure consisted of relatively coarse acicular alpha in a matrix of dehydrogenated transformation product (Figure 3(b)). Above 0.85 wt pct hydrogen yield strength appears to be independent of previous hydrogen content, dependent only upon transformation and dehydrogenation. The microstructures of these specimens showed only the dehydrogenated transformation product of Figure 4. Elongation varied slightly with previous hydrogen content, decreasing with increasing hydrogen. This effect was more pronounced at the lowest dehydrogenation temperature. Analyses for hydrogen after testing indicated the lower ductilities were not associated with residual hydrogen; for all specimens analyzed, final hydrogen content was less than 10 ppm. Tensile results for specimens cooled to room temperature between beta treatment and transformation are

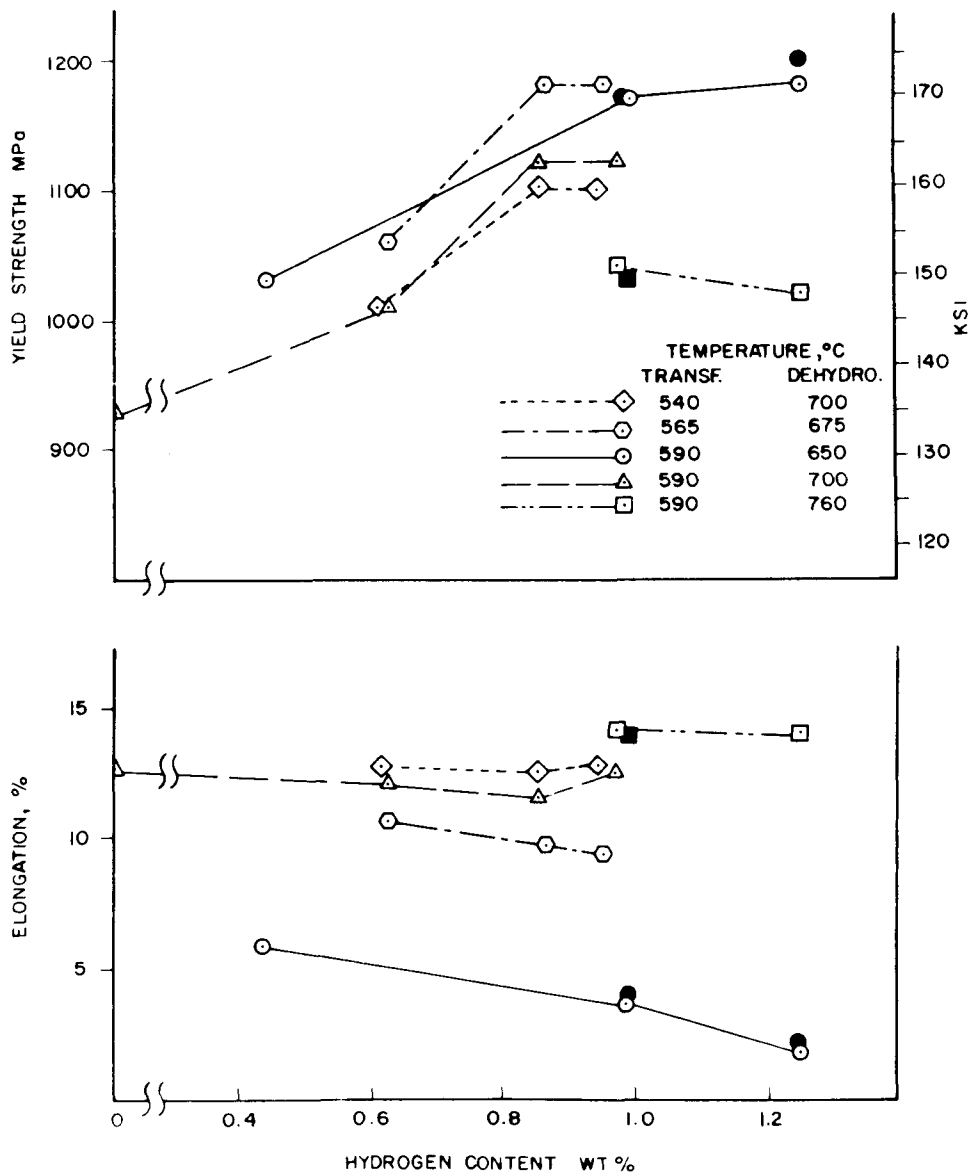


Fig. 5—Yield strength (0.2 pct offset) and elongation (pct in 38 mm) as a function of hydrogen content for various transformation and dehydrogenation treatments. Solid symbols indicate specimens cooled to room temperature between beta treatment and transformation.

not significantly different from the others. This result is consistent with the similarity in microstructure for these two methods of transformation.

In Figure 6, yield strength and elongation are plotted as a function of dehydrogenation temperature for a constant hydrogen content, nominally 1.0 wt pct, for various transformation temperatures. Within the range 540 °C to 590 °C, where as-transformed microstructures were similar, transformation temperature has no significant effect on these properties. As dehydrogenation temperature and hence the size and spacing of the beta particles increase (see Figure 4), the yield strength decreases and the elongation increases. For specimens transformed at 700 °C, the as-transformed microstructure appears to be controlling and the properties do not change significantly with dehydrogenation temperatures. Presumably, the presence of the acicular primary

alpha, Figure 3(c), decreases both strength and ductility regardless of the microstructure of the matrix.

Comparison of tensile properties and microstructure indicates that the highest yield strength materials are characterized by very fine, equiaxed beta particles in a matrix of essentially equiaxed alpha. This microstructure is produced in material in which the hydrogen content was sufficient (>0.8 pct) to produce transformation product without proeutectoid alpha and in which transformation temperature was low enough to produce finely divided eutectoid structure (540 °C to 590 °C). Lower dehydrogenation temperatures (650 °C to 700 °C) remove hydrogen without growth of particles and concurrent decrease in strength. Within these ranges of hydrogen content and transformation and dehydrogenation temperatures, difference in tensile properties could not be related to microstructural differences



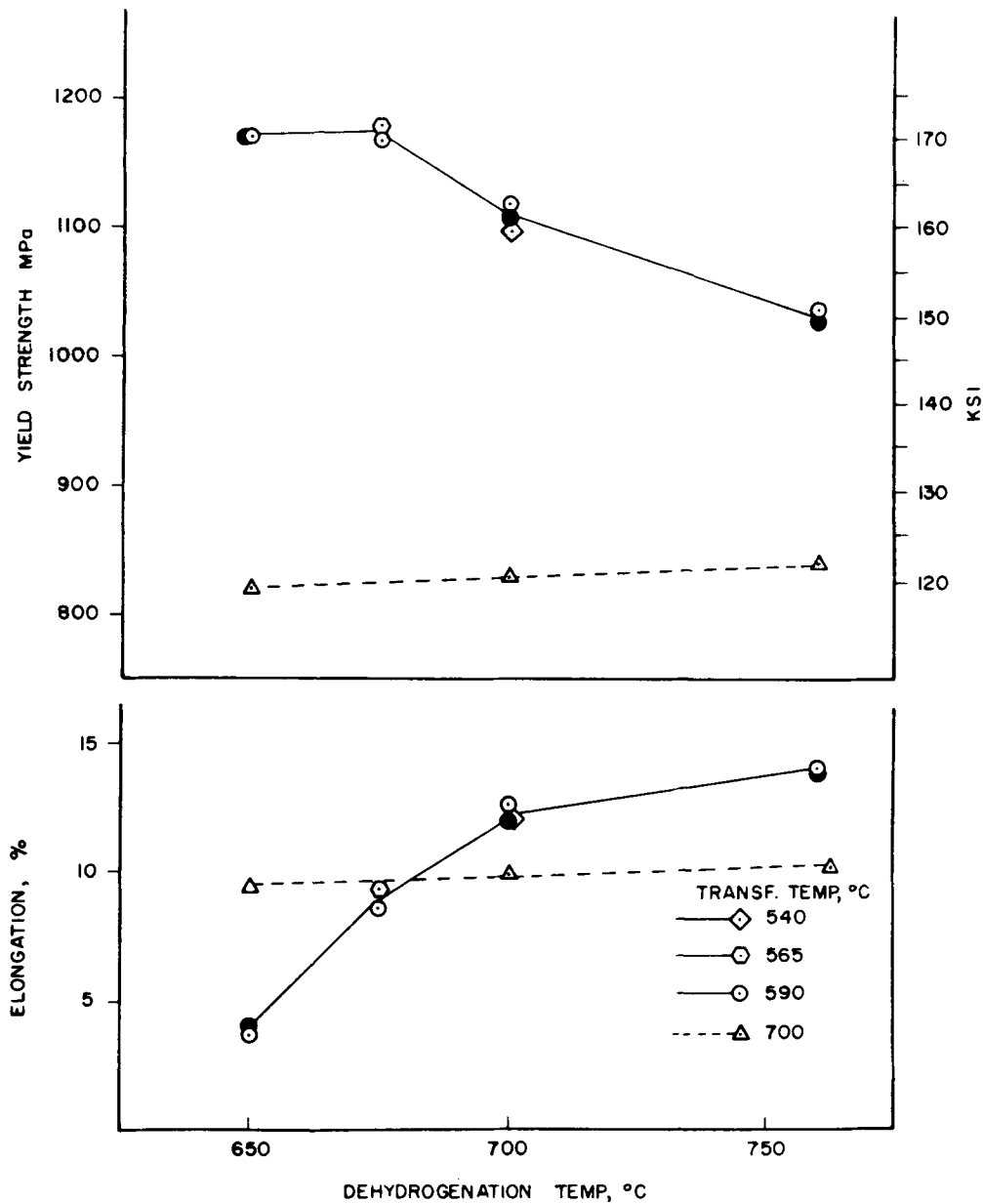


Fig. 6—Yield strength (0.2 pct offset) and elongation (pct in 38 mm) of Ti-6Al-4V-1H specimens as a function of dehydrogenation temperatures. Solid symbols indicate specimens cooled to room temperature between beta treatment and transformation.

observed by scanning electron microscopy. The highest yield strengths were obtained at the lowest dehydrogenation temperature, 650 °C, but were achieved with considerable loss of ductility. Equivalent strength with reasonable ductility was achieved by dehydrogenating at 675 °C.

Such property differences can be related to dislocation density and substructure. The addition of hydrogen to Ti-6Al-4V significantly decreases the density. Careful measurement of cylindrical specimens before and after hydrogenation indicates that for 1.0 wt pct hydrogen this change is approximately 5 pct which is comparable to the 5.35 pct change determined from beta lattice parameter measurements.<sup>13</sup> Dehydrogenation results in a corresponding increase in density, but the strains produced in the lattice result in a fine substructure and an array of dislocations as shown in the transmission electron micrograph of Figure 7.

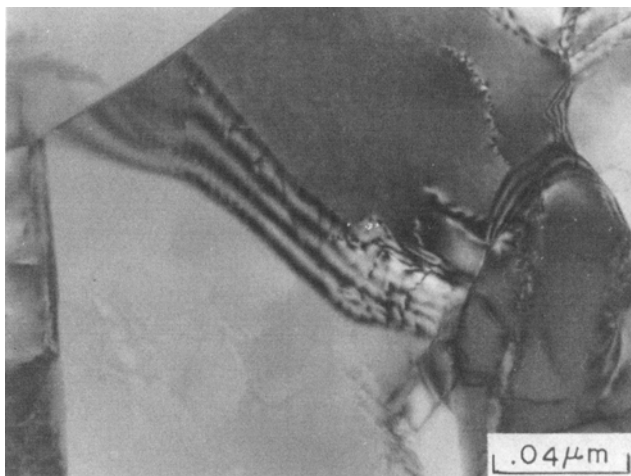
For a specimen dehydrogenated at 650 °C (Figure 7(a)) the structure is that of a heavily cold-worked material. As expected, this material had high yield strength and low ductility. At higher temperatures (Figure 7(b), 760 °C) some of the dislocations are annealed out and some substructure growth occurs, resulting in improved ductility and lower yield strength. The best combination of strength and ductility was obtained for specimens dehydrogenated at 675 °C having the substructure shown in Figure 7(c).

#### IV. CONCLUSIONS

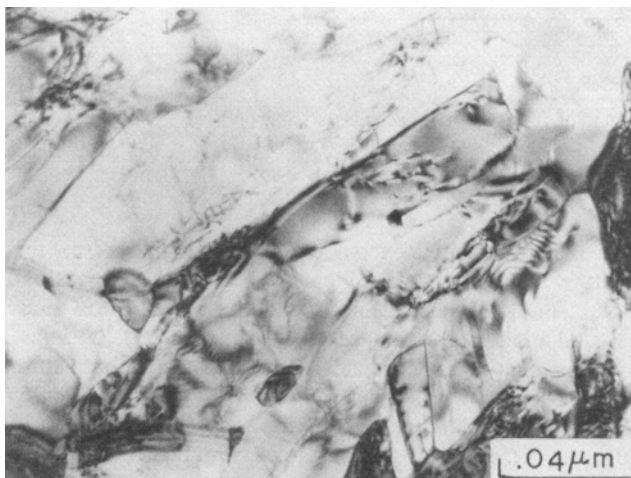
1. Significant variations in the microstructure and tensile properties of Ti-6Al-4V can be achieved by charging with hydrogen, heat treating, and subsequently dehydrogenating.



(a)



(b)



(c)

Fig. 7—Transmission electron micrograph of Ti-6Al-4V-1H: (a) heat treated as for Fig. 4(a). Note small subgrains and large number of dislocations. (b) Heat treated as for Fig. 4(c). Note relatively large subgrains and few dislocations. (c) Specimen which showed highest yield strength and good ductility. Subgrain size and dislocation density are intermediate between (a) and (b).

2. The major variations in microstructure are the result of a solid-solid phase transformation resulting in three finely divided phases: alpha, beta, and hydride. The extent of this transformation and the resultant microstructures vary with hydrogen content and transformation temperature.
3. Additional variations in microstructure and in substructure occur as a result of dehydrogenation. The temperature of dehydrogenation determines the size and spacing of the alpha and beta phases which exist in the final product and influences dislocation density and subgrain size.
4. The highest tensile strength and reasonable ductility are achieved when the transformation/dehydrogenation cycle produces a fine dispersion of equiaxed beta in a matrix of alpha, also essentially equiaxed and a reasonable density of dislocations produced by density changes during hydrogenation and dehydrogenation.
5. Low strength and ductility occur when hydrogen content and heat treatment result in an acicular alpha phase.

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