CHEMISTRY AND HEAT-TREATMENT EFFECTS
ON MECHANICAL AND MICROSTRUCTURAL
PROPERTIES OF HEAT-TREATED, BETA-EXTRUDED
Ti-6Al-6V-2Sn

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Published April 1979

Topical Report.

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STRUCTURAL PROPERTIES OF HEAT-TREATED, BETA-EXTRUDED Ti-6Al-6V-2Sn

BDX-613-2170 (Rev.), Topical Report, Published April 1979

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The mechanical behavior of beta-extruded Ti-6Al-6V-2Sn was examined after a variety of sub-transus heat treatments. The microstructural variations resulting from the range of heat treatments studied also were examined. A range of alloy chemistries, within commercial limits, was used to evaluate the effect of this variable on mechanical properties. The strength-toughness combinations obtained in beta-extruded Ti-6Al-6V-2Sn ranged from about 895 MPa and $82.5 \text{ MPa}\sqrt{\text{m}}$ for duplex annealed material to 1200 MPa and $54.9 \text{ MPa}\sqrt{\text{m}}$ for solution treated and peak aged material. Chemistry variations had less effect on mechanical properties than would have been the case with alpha-beta processing.

FOA:TR/7

Originally published as BDX-613-1812 on April 4, 1977.

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A prime contractor with the United States
Department of Energy under Contract Number
DE-AC04-76-DP00613

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SUMMARY

The mechanical behavior of beta-extruded Ti-6Al-6V-2Sn has been examined after a variety of sub-transus heat treatments. Increasing strength has been found to result in a gradual decrease in fracture toughness. The fracture toughness that may be achieved at any given strength level is also a function of prior solution treatment. Optimum toughness at intermediate strength levels (yield strength of approximately 1035 MPa) is associated with lower solution treatment temperatures, while at high strength levels (yield strength of approximately 1200 MPa), the use of higher solution temperatures appears to enhance the maximum achievable toughness. Comparison of the present results with those previously obtained for alpha-beta processed Ti-6Al-6V-2Sn also indicates that at any given strength level, beta processing will result in a higher fracture toughness and a somewhat lower tensile ductility.

Although the results reported for the solution treated and aged material were all determined on a single alloy chemistry, similar mechanical properties tests were carried out on duplex annealed materials of this original and seven different alloy chemistries. All these chemistries were within the commercial limits of Ti-6Al-6V-2Sn alloy. The results of these tests show that alloy chemistry variations have little or no effect on the tensile strength and only a slight effect on the fracture toughness of this beta processed material. These results are in contrast to what would be expected for recrystallized annealed alpha-beta processed material.

Finally, microscopic evidence suggests that the ordering in the primary alpha phase which occurs after high temperature aging does not play a predominant role in limiting the fracture toughness of high strength alpha-beta titanium alloys; indeed, the deformation limiting characteristics of the interface alpha phase may be the principal factor controlling the toughness of these alloys.

DISCUSSION

SCOPE AND PURPOSE

Relatively little data are available on the mechanical properties of beta-processed Ti-6Al-6V-2Sn alloy. Therefore, Bendix conducted this study for Sandia Laboratories, Livermore, to evaluate the mechanical properties, tensile strength, and fracture toughness of beta-extruded Ti-6Al-6V-2Sn in a wide range of heat treat conditions. Sandia Laboratories, Albuquerque carried out a microstructural characterization of this material in the as-extruded and in several of the heat treat conditions. Pertinent parts of these results are included in this report.

This joint Bendix and Sandia Laboratories project has resulted in the characterization of the mechanical and microstructural properties of heat-treated, beta-extruded Ti-6Al-6V-2Sn alloy and in the determination of the effect of chemistry variations, within commercial limits, on these mechanical properties. The alloy was examined in both the duplex annealed (DA) and the solution treated and aged (STA) conditions. Because a fairly wide range of alloy element concentration is allowed by MIL-T-9047F, the possible effect of these variations on mechanical properties was made a part of this study.

PRIOR WORK

Although Ti-6Al-6V-2Sn and Ti-6Al-4V alloys are similar, age hardenable, alpha-beta titanium alloys, the former is generally thought to possess greater through-thickness hardenability. This increased hardenability is typically associated with more uniform mechanical behavior in heavier section sizes. Prior investigations¹⁻⁹ have shown that processing variations can have a marked influence on these mechanical properties. Procedures which involve either beta working or heat treatments above the beta transus temperature can result, at a given strength level, in improving fracture toughness and stress corrosion cracking resistance.^{3,5,10-16} These improvements, when compared to those obtainable by wholly alpha-beta processing, are achieved at the expense of a slight reduction in the maximum strength achievable and a more significant reduction in tensile ductility. Previous studies of the mechanical properties of beta processed titanium alloys have been concentrated on Ti-6Al-4V¹⁰⁻¹⁵ with very little attention being given to other alloy systems.⁵ Likewise, there is little or no information available in the literature on the effects of chemistry variations on mechanical properties of Ti-6Al-6V-2Sn.

ACTIVITY

Experimental Procedure

The chemical analysis of the Ti-6Al-6V-2Sn evaluated in this study is given in Table 1. The majority of this work was carried out on heat number 1. However, heats 2 through 8 were used for the alloy chemistry variations study. All heats used in this study were alpha-beta blocked at 900°C and beta extruded at 985°C. This procedure resulted in a final extrusion ratio of 4.4:1. All test specimens were removed from 140- to 152-mm long prolongations taken from predetermined locations along the extrusion length (Figure 1). Figure 2 shows the relationship between the ring prolongation and the various specimen orientations examined.

Several heat treatment conditions have been investigated. Solution treated and aged specimens, machined from 32-mm arcs of the ring prolongations, were solution treated 1 hour; water quenched; aged at 400, 450, 500, 550, 600, 650, and 700°C for specified times; and then air cooled. Four separate schedules were used:

Treat at 850°C and age 4 hours at each temperature;

Treat at 850°C and age 8 hours at each temperature;

Treat at 900°C and age 4 hours at each temperature; and

Treat at 900°C and age 8 hours at each temperature.

Duplex annealed specimens were prepared from 152-mm arcs. These specimens were annealed 2 hours at 900°C, air cooled, annealed another 2 hours at 785°C, and air cooled again. Isothermal annealed specimens were solution treated 1 hour at 850°C and quenched directly, some to 600°C for 8 hours and others to 500°C for 8 hours.

Evaluation of the mechanical behavior of the beta-extruded Ti-6Al-6V-2Sn included both tensile and fracture toughness measurements. Tensile test specimens were tested according to ASTM Tension Testing of Metallic Materials (E8-69). The rate of strain was maintained at 0.005 per minute through the 0.2 percent offset yield strength and then increased to produce failure in approximately one additional minute. The fracture toughness measurements included both compact tension specimens tested according to ASTM Plane-Strain Fracture Toughness of Metallic Materials (E399-74) and instrumented, precracked Charpy specimens which had been machined according to ASTM Notched Bar Impact Testing of Metallic Materials (E-23-72). For the latter, the Charpy V-notch was initially sharpened by electrical discharge machining and then precracked; the final crack depth to specimen width ratio (a/w) for these specimens was between 0.45 and 0.55.

Table 1. Chemical Analysis of Ti-6Al-6V-2Sn Alloys

Element	Heat							
	1	2	3	4	5	6	7	8
Weight Percent								
Al	5.7	5.3	5.9	5.9	5.6	5.7	5.5	5.4
O ₂	0.155	0.126	0.179	0.121	0.185	0.149	0.160	0.158
Cu	0.77	0.46	0.50	0.82	0.79	0.73	0.66	0.61
V	5.7	5.8	5.9	5.9	5.9	6.0	5.4	5.8
Sn	2.1	2.0	2.0	2.1	2.0	2.2	2.1	2.2
Fe	0.75	0.65	0.68	0.72	0.69	0.73	0.66	0.47
N ₂	0.011	0.010	0.008	0.010	0.008	0.009	0.007	0.009
C	0.02	0.02	0.02	0.02	0.02	0.03	0.02	0.03
H ₂ (ppm)	45*	26*	25	22	23	29	30	29

*Hydrogen concentration, determined on this heat only, actually varied from 30 ppm on the inside diameter to 60 ppm on the outside diameter.

Figure 1. Relationship of Piece Parts and Ring Prolongations to Beta-Extruded Pipe With Ring Prolongation Dimensions

Charpy type specimens were tested at an impact velocity of 1 m/s which is equivalent to 11.1 J of energy at impact. All specimen orientation nomenclature corresponds to that given in ASTM E399-74.

Because of the very large number of similar heat treatments and tests, some conventions of nomenclature should be established. First is a shorthand notation for the solution treated and aged material. These heat treatments will be designated according to the following example: 850/450/8, where the first number is the solutionizing temperature, the second number is the aging temperature, and the third number is the aging time in hours. All temperatures are given in degrees Celcius. Solutionizing is for one hour followed by water quenching, and aging is followed by air cooling. Second, Charpy testing will correspond to tests from which dynamic fracture toughness, K_{Id} , was calculated. Finally, fracture toughness testing will be used to designate compact tension specimens from which static fracture toughness, K_{Ic} , was calculated.

Figure 2. Orientation and Nomenclature of Specimens Within Ring Prolongations

The microstructural characterization portion of this study generally used material removed from previously failed mechanical test specimens. Optical metallography used standard techniques with a final etchant of 1.5 volume percent HF, 3.5 volume percent HNO_3 , and the balance H_2O . The measurements of the volume fraction primary alpha, the prior beta grain size, and the grain boundary primary alpha thickness were performed using quantitative metallographic procedures.¹⁷ The microstructures observed were essentially equiaxed; therefore, no differentiation will be made between the measurements obtained on the three orthogonal surfaces (longitudinal and long and short transverse).

Texture determination was in accordance with ASTM Preparing Quantitative Pole Figures of Metals (E81-63 Reapproved 1974) with $\text{CuK}\alpha$ radiation and a nickel filter. The X-ray pole figures were obtained by the Schulz technique on a texture goniometer, with background corrections performed prior to automatic plotting of the pole figure.¹⁸

Finally, transmission and scanning electron microscopy observations were made using an electron microscope operated at 200 kV and a scanning electron microscope operated at 25 kV, respectively. Transmission samples were prepared as detailed elsewhere.¹⁹ It is to be noted that all fractographic examinations were normal to the original fracture surface.

Textural Analysis

Figure 3 illustrates a (0002) α pole figure obtained from the central region of the extrusion. Each contour line in this figure represents an intensity increment of 10. Unfortunately, no quantitative comparison with a random sample can be made because none was available at the time the measurements were made. A qualitative comparison of this figure with those obtained in other investigations²⁰ does, however, indicate that it is typical of alpha-beta titanium alloys which have been beta worked and transformed from the beta to alpha phase upon cooling; that is, a concentration of basal planes is observed to be at an angle of approximately 40 degrees from both the long transverse direction and the sample surface.

Optical and Transmission Microscopy

As-Extruded Condition

Optical examination (Figure 4) of the as-extruded condition revealed an essentially equiaxed beta grain structure with a prior beta grain size and primary alpha volume fraction of 216.3 μm and 0.49, respectively. The prior beta grain boundaries also contained an almost continuous layer of primary alpha, the latter having a mean thickness of 4.7 μm . In addition, isolated areas were observed wherein the normally acicular primary alpha tended to be rounded (see arrow in Figure 4). Electron microprobe examination of these areas showed that the primary alpha contained within these regions, as well as the primary alpha located at the prior beta grain boundaries, did not contain as high a concentration of vanadium and iron when compared with the more general Widmanstätten (alpha plus beta) matrix. This observation suggests that both these regions and the grain boundary primary alpha layer form first when the extrusion is cooled from the beta extrusion temperature. Further examination showed that the aforementioned regions of localized concentration differences were not materially affected by any of the subsequent solution or duplex annealing treatments.

Transmission electron microscopy of the as-extruded condition shows Widmanstätten alpha separated by the beta matrix (Figure 5). Dark field microscopy (Figure 6), taken using the interface alpha reflection, showed that the alpha-beta interface contained the

LONGITUDINAL

Figure 3. Pole Figure of As-Beta-Extruded Ti-6Al-6V-2Sn (Material examined was from center of extrusion thickness; note the concentration of basal planes at angle of 40 degrees to both long transverse direction and sample surface.)

interface phase recently reported by Rhodes and Williams.²¹ The interface region contains a high dislocation density and a high density of alpha phase particles. Figure 6 clearly shows that the interface alpha phase is not continuous; it appears as discrete particles lying along the alpha-beta interface. The interface phase was observed through all the microstructural conditions examined during this study and seemed to be relatively unaffected by subsequent heat treatment.

Finally the primary alpha phase itself had a low dislocation density and the transformed beta matrix contained some secondary alpha particles.

Figure 4. Optical Photomicrograph of As-Beta-Extruded Ti-6Al-6V-2Sn (Structure consists of equiaxed prior beta grains outlined by continuous layer of primary alpha. The normally acicular primary alpha within the prior beta grains contains isolated areas where the primary alpha tends to be rounded, as shown at arrow in lower left corner.)

Duplex Annealed

Figure 7 shows the microstructure after each stage of the two-stage duplex annealing heat treatment. A comparison with the as-extruded conditions (Figure 4) suggests that some rounding of the acicular primary alpha phase has occurred during the annealing cycle. In addition, the quantitative measurements indicate that while the

Figure 5. Transmission Electron Photomicrograph of As-Beta Extruded Ti-6Al-6V-2Sn, Showing Widmanstatten Alpha Separated by the Beta Matrix

prior beta grain size and volume fraction primary alpha have increased slightly to 248 μm and 0.52 from 216.3 μm and 0.49, respectively, the grain boundary alpha thickness has decreased to 3.6 μm from 4.7 μm .

Transmission electron microscopy indicates that the primary alpha phase in the duplex annealed condition again contained a low dislocation density which was, however, generally arranged in planar arrays (Figure 8). In addition, further evidence for increased formation of secondary alpha in the beta matrix was obtained.

Figure 6. Dark Field Transmission Photomicrograph of As-Beta-Extruded Ti-6Al-6V-2Sn (Photo taken using the interface alpha reflection showing the high density of discrete alpha phase interface particles as well as the high dislocation density at the interface.)

All data reported here were taken on material from heat 1. However, there is no reason to believe that similar analysis carried out on heats 2 through 8 would yield substantially different results.

Solution Treated and Aged

Several solution treated and aged (STA) conditions were examined. An optical micrograph following solution treatment at 850°C is presented in Figure 9. Transmission electron microscopy showed

a. AFTER 2 HOURS AT 900°C, AIR COOLED

b. AFTER ADDITIONAL 2 HOURS AT 785°C, AIR COOLED

Figure 7. Optical Photomicrographs of Beta-Extruded Ti-6Al-6V-2Sn After Each Phase of Duplex Anneal (This heat treatment results in a rounding of the acicular primary alpha, an increase in prior beta grain size and volume fraction primary alpha, and a decrease in grain boundary alpha thickness compared to the as-beta-extruded condition.)

Figure 8. Transmission Electron Photomicrograph of Duplex Annealed Ti-6Al-6V-2Sn, Showing Planar Arrays of Dislocations Resulting From This Heat Treatment

that the structure consisted of primary alpha and a heavily dislocated martensitically transformed beta matrix. Quantitative measurements indicate that the 850°C solution treatment increased the amount of primary alpha to 75 volume percent without materially affecting either the prior beta grain size or the grain boundary alpha thickness. A more detailed examination of the effect of the 900°C solution treatment on the prior beta grain size, the

Figure 9. Optical Photomicrograph of Ti-6Al-6V-2Sn, Solution Treated 1 Hour at 850°C and Water Quenched

volume fraction alpha, and the grain boundary alpha thickness is presently underway at Sandia Laboratories, Albuquerque, and will be reported at a later date.

Probably the most significant feature observed in the examination of the various solution treated and aged condition was the presence of Ti_3Al in the primary alpha phase after the 900/650/8 treatment. Its appearance is best illustrated by the $(0002)\alpha$ selected area diffraction pattern presented in Figure 10, where the superlattice reflections correspond to the DO_{19} lattice of Ti_3Al .

Finally, aging of solution treated Ti-6Al-6V-2Sn results in precipitation of alpha in the previously transformed beta matrix.

Figure 10. Selected Area Diffraction of Ti-6Al-6V-2Sn (This sample was solutionized 1 hour at 900°C, water quenched, aged 8 hours at 650°C, and air cooled. The zone axis is the (0002) α and the superlattice reflections correspond to the DO₁₉ lattice of Ti₃Al.)

To date no difference in alpha phase precipitation has been observed between material solutionized at 850°C and at 900°C prior to aging. A recent paper²² does suggest that changing the solution temperature from 800 to 900°C should result in a morphological transition from needle or platelike alpha to lamellar alpha precipitates upon aging. This observation was, however, made after much longer aging times than the maximum of 8 hours used in this study. Consequently, it is not surprising that this transition was not observed in the present study.

Mechanical Properties

Tensile Behavior

Results of the yield and ultimate tensile strength tests for all solution treated and aged conditions of heat 1 are given in Table 2. Percent elongation in 25.4 mm and percent reduction in area (RA) results are also presented in this table. Figure 11 shows the longitudinal yield strength versus aging temperature for all heat treat conditions. The data show that STA Ti-6Al-6V-2Sn exhibits the characteristics of classical age hardening behavior. That is, the strength increases with aging temperature up to a maximum value and then decreases as aging temperature is further increased. Note that the material solutionized at 900°C appears to reach the peak strength condition when aged at 450°C. This is true for aging times of either 4 or 8 hours. On the other hand, the material solutionized at 850°C exhibits overaging behavior even when aged at 400°C for 4 hours.

Ultimate tensile strength as a function of aging temperature is plotted in Figure 12. The ultimate strength followed essentially the same trend as the 0.2 percent yield strength and showed expected peak aging and overaging behavior.

The beta processing of both Ti-6Al-4V and Ti-6Al-6V-2Sn alloys has previously been shown to noticeably reduce the tensile elongation as compared to conventional alpha beta processed material.¹⁶ Figure 13 shows a plot of percent elongation in 25.4 mm as a function of yield strength. Note that the elongation appears to be dependent only upon the strength level and is not a function of the details of heat treatment, whether duplex annealed or solution treated and aged. These values of tensile elongation are about 4 to 8 percent lower than those for Ti-6Al-6V-2Sn heat treated to the same strength level but processed in the alpha-beta temperature range.²³ Percent reduction in area as a function of yield strength is shown in Figure 14. Again, the details of the heat treatment do not appear to influence the percent reduction in area at a given strength level. The strength level appears to be the sole determining factor. Reduction in area of beta processed Ti-6Al-6V-2Sn is only about half that for alpha-beta processed material at the same strength level.²³

Although yield strength levels above 1240 MPa can be achieved in beta processed Ti-6Al-6V-2Sn (see, for example, heat treatments 900/400/8, 900/450/8, 900/450/4, and 900/500/4), these conditions result in tensile elongations of less than 4 percent. To achieve a normally acceptable elongation of 10 percent, heat treatments such as 800/600/8 and 900/650/8 must be used. These heat treatments produce yield strengths in the 1034 and 1066 MPa range. These

Table 2. Tensile Test Results for Solution Treated and Aged Longitudinal Specimens

Aging Temperature (°C)	0.2 Percent Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation in 25.4 mm (Percent)	Reduction in Area (Percent)
Schedule 1*				
400	1178	1316	1.3	1.1
450	1171	1357	1.1	1.7
500	1158	1295	7.3	6.1
550	1116	1213	9.3	12.3
600	1040	1137	10.9	19.6
650	978	1075	13.5	25.1
700	923	1027	15.2	22.4
Schedule 2				
400	1178	1302	1.9	2.0
450	1171	1344	3.3	1.8
500	1151	1268	6.0	8.7
550	1123	1220	8.4	11.2
600	1047	1144	10.9	17.0
650	951	1054	15.4	27.4
700	882	1006	16.0	26.1
Schedule 3				
400	1226	1137	1.7	2.8
450	1302	1371	2.1	3.3
500	1261	1364	2.1	4.6
550	1185	1288	4.0	6.7
600	1123	1199	7.5	10.7
650	1082	1192	10.1	13.7
700	1013	1116	11.1	19.8

Table 2 Continued. Tensile Test Results for Solution Treated and Aged Longitudinal Specimens

Aging Temperature (°C)	0.2 Percent Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation in 25.4 mm (Percent)	Reduction in Area (Percent)
Schedule 4				
400	1240	1412	3.6	3.3
450	1247	1378	2.6	2.4
500	1233	1344	3.4	4.7
550	1192	1261	4.6	6.3
600	1130	1226	6.6	11.1
650	1054	1158	11.0	15.7
700	944	1047	14.6	25.4

*Schedules:

1. Solution treated 1 hour at 850°C, water quenched, aged 4 hours, and air cooled.
2. Same as 1, except aged 8 hours.
3. Same as 1, except solution treated 1 hour at 900°C.
4. Same as 3, except aged 8 hours.

values can be compared to the alpha-beta processed material reported by Lewis and others²³ which had a yield strength of 1160 MPa and 14 percent elongation.

Dynamic Fracture Toughness

Figure 15 shows the dynamic fracture toughness, K_{Id} , determined from precracked Charpy impact tests, as a function of aging temperature. All K_{Id} values were calculated using linear elastic fracture mechanics, as there was no evidence of general yielding before fracture. The values plotted in Figure 15 are taken from Table 3. As expected, toughness increases with increasing aging temperature (that is, with decreasing tensile strength). However, this figure does indicate that, contrary to the tensile behavior, the details of the heat treat procedure do have an effect on toughness. Solution treatment at 850°C generally produced a

DATA ARE AVERAGE OF TWO TESTS

Figure 11. Yield Strength Versus Aging Temperature for all Heat Treat Conditions Studied

higher fracture toughness at a given aging temperature above 450°C than did solutionizing at 900°C. In other words, the lower strength STA material developed higher toughness when solutionized at 850°C and the higher strength material developed higher toughness when solutionized at 900°C. In addition, aging for only 4 hours following solutionizing at 850°C further enhanced toughness. The causes for these differences, as yet undetermined, form the basis for further studies.

The interrelationship between yield strength and dynamic fracture toughness is represented in Figure 16. This figure shows that an acceptable combination of yield strength and toughness can be realized from heat-treated, beta-processed Ti-6Al-6V-2Sn. These results again indicate that, for yield strengths between 827 and

DATA ARE AVERAGE OF TWO TESTS

Figure 12. Ultimate Tensile Strength Versus Aging Temperature for all Heat Treat Conditions Studied

1034 MPa, maximum fracture toughness can be achieved by solutionizing at 850°C and aging for 4 hours. It also suggests that, at strength levels above 1170 MPa, solutionizing at 900°C results in maximum toughness levels.

Tensile and Fracture Toughness Behavior of Selected Heat Treatments

Based on the above results, two solution treated and aged conditions were chosen to produce yield stresses of 1034 to 1068 MPa and fracture toughness of approximately 55 MPa√m. Details of these heat treat cycles and the test results are given in Table 4. In addition to the preliminary tests already discussed for these heat treatments, tensile tests in both the long and short transverse

DATA ARE AVERAGE OF THREE TESTS

Figure 15. Dynamic Fracture Toughness Versus Aging Temperature for all Heat Treat Conditions Studied

there is a substantial loss in toughness (30 to 40 percent) in going from the 1034 MPa to the 1170 to 1240 MPa strength level. Also, the tensile ductilities resulting from these heat treatments are considerably lower than the generally preferred 10 percent minimum elongation.

Note from Table 5 that the K_{Ic} and K_{Id} data show good agreement for the high strength heat treatment, but not as good as for the 1068 MPa heat treatment (Table 4). This effect is shown graphically in Figure 16 for those heat treatments selected for further study and for the DA material where the symbols with the numbers inside

Table 3. Fracture Toughness Test Results
for LT Orientation Solution
Treated and Aged Specimens

Aging Temperature (°C)	Fracture Toughness (MPa√m)	
	K _{Id}	K _{Ic}
Schedule 1*		
400	35.1	
450	43.9	
500	56.9	
550	67.1	
600	74.8	
650	83.2	
700	87.2	
Schedule 2		
400	36.5	35.9
450	40.6	38.7
500	53.7	48.2
550	60.9	60.3
600	68.3	67.5
650	76.2	81.6
700	79.4	89.3
Schedule 3		
400	48.2	
450	44.7	
500	48.5	
550	52.3	
600	58.1	
650	66.9	
700	70.6	

Table 3 Continued. Fracture Toughness Test
Results for LT
Orientation Solution
Treated and Aged
Specimens

Aging Temperature (°C)	Fracture Toughness (MPa√m)	
	K _{Id}	K _{Ic}
Schedule 4		
400	48.2	
450	47.3	
500	53.4	39.8
550	54.9	
600	70.5	
650	66.9	65.5
700	74.5	

*For schedule treatments, see note to
Table 2.

show K_{Ic} values from LT orientation compact tension specimens. The fairly good agreement between K_{Ic} and K_{Id} values has been noted but should not be given undue weight at this time. This relationship is being investigated in more detail.

Mechanical Testing of Duplex Annealed Material

Witness pieces from three duplex annealed extrusions from heat 1 (Table 1) were tested for longitudinal and long and short transverse tensile properties, LT orientation precracked Charpy impact properties, and LT orientation fracture toughness. The test specimens were prepared from 57-mm-thick sections. The averaged results given in Table 6 under Heat 1 show there is little difference in tensile properties with test direction. The duplex anneal condition resulted in the lowest strength, highest toughness material evaluated in this study.

Figure 16. Yield Strength Versus Dynamic Fracture Toughness for all Heat Treat Conditions Studied

In addition to the tests performed on material from heat 1, a similar series of tests were performed on duplex annealed material from seven additional and different alloy chemistries of Ti-6Al-6V-2Sn (Table 1). The actual chemistries of all alloys lie within the compositional limits of MIL-T-9047, so these variations in composition could well be expected during the production lifetime of any product fabricated from Ti-6Al-6V-2Sn. The total spread in yield strength for longitudinal specimens (Table 6) ranges from 1077 MPa for heat 2 to 1130 MPa for heat 5. This corresponds to only 5 percent spread in these values. Because the expected error in tensile tests is 2 percent or greater, these tensile test results indicate a minimal effect of chemistry on the yield strength of this beta-extruded and duplex annealed material.

Table 4. Results of Additional Tests of Specimens Heat Treated to Give 1068 MPa Yield Strength

	Specimen Orientation					
	Tensile Tests*			Fracture Toughness Tests		
	Longitudinal	Transverse Long	Short	LT	TL	SL
Solution treated 1 hr at 850°C, water quenched, aged 8 hr at 600°C, air cooled						
0.2 Percent offset Yield Strength (MPa)	1047	1068	1094			
Ultimate Tensile Strength (MPa)	1144	1158	1164			
Elongation in 25.4 mm (Percent)	10.9	8.8	6.0			
Reduction in Area (Percent)	17.0	14.2	14.2			
Fracture Toughness (MPa√m)						
K_{Id}^{**}				68.3		
K_{Ic}^*				66.6	66.2	61.1
Solution treated 1 hr at 900°C, water quenched, aged 8 hr at 650°C, air cooled						
0.2 Percent Offset Yield Strength (MPa)	1054	1068	1068			

Table 4 Continued. Results of Additional Tests of Specimens Heat Treated to Give 1068 MPa Yield Strength

	Specimen Orientation					
	Tensile Tests*			Fracture Toughness Tests		
	Longitudinal	Long	Short	LT	TL	SL
Ultimate Tensile Strength (MPa)	1158	1164	1164			
Elongation in 25.4 mm (Percent)	11.0	10.5	7.0			
Reduction in Area (Percent)	15.7	15.7	12.6			
Fracture Toughness (MPa√m)						
K_{Id}^{**}				66.9		
K_{Ic}^*				65.5	61.1	57.6
*Average of two tests						
**Average of three tests						

Table 5. Results of Additional Tests of Specimens Heat Treated to Give Maximum Useable Strength

	Longitudinal Orientation Tensile Tests		LT Orientation Fracture Toughness Tests	
	Specimens*		Specimens*	
	A	B	A	B
0.2 Percent Offset Yield Strength (MPa)	1151	1233		
Ultimate Tensile Strength (MPa)	1268	1344		
Elongation in 25.4 mm (Percent)	6.0	3.4		
Reduction in Area (Percent)	8.7	4.7		
K_{Id} (MPa $\sqrt{m}$)			53.7	53.4
K_{Ic} (MPa $\sqrt{m}$)			48.3	39.8

*Specimen A was solution treated 1 hr at 850°C, water quenched, aged 8 hr at 500°C, and air cooled; specimen B was solution treated 1 hr at 900°C, water quenched, aged 8 hr at 500°C, and air cooled. Tensile and K_{Ic} results are average of two tests, and K_{Id} results are average of three tests.

On the other hand, there is a considerable spread in the fracture toughness values determined from compact tension specimens. These values ranged from 90.9 MPa $\sqrt{m}$ for heat 1 to 106.9 MPa $\sqrt{m}$ for heats 2 and 4. This corresponds to a spread of 16 percent, considerably greater than the 5 percent spread which can be expected for fracture toughness testing. Kennedy²⁴ has made predictions of the effects of variations of aluminum and oxygen contents on the yield strength and fracture toughness of alpha-beta processed and recrystallized annealed Ti-6Al-6V-2Sn. These predicted changes and the measured variations for heats 2 through 8 versus heat 1 are shown in Table 7. It is evident from this table that chemistry variations in beta-extruded and duplex

Table 6. Tensile and Toughness Test Results (Averages of Two Tests) for Duplex Annealed Specimens of all Chemistries

Specimen Orientation	Tensile Tests				Fracture Toughness (MPa√m)	
	0.2 Percent Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation in 25.4 mm (Percent)	Reduction in Area (Percent)	K _{Id}	K _{Ic}
Heat 1						
Longitudinal	861	971	17.9	28.9	98.5	90.9
Long Transverse	875	971	17.0	28.9		
Short Transverse	868	971	20.5	25.7		
LT						
Heat 2						
Longitudinal	1077	1228	16.9	27.2	55.1	106.9
Long Transverse	1086	1228	16.5	29.5		
LT						
TL						
Heat 3						
Longitudinal	1121	1282	19.1	32.1	85.1	95.5
Long Transverse	1121	1282	17.9	35.6		
LT						
TL						

Table 6 Continued. Tensile and Toughness Test Results (Averages of Two Tests)
for Duplex Annealed Specimens of all Chemistries

Specimen Orientation	Tensile Tests				Fracture Toughness (MPa√m)	
	0.2 Percent Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation in 25.4 mm (Percent)	Reduction in Area (Percent)	K _{Id}	K _{Ic}
Heat 4						
Longitudinal	842	1237	17.0	27.0		
Long Transverse	1068	1255	16.8	28.3		
LT						106.9
TL						107.4
Heat 5						
Longitudinal	1130	1291	16.0	28.4		
Long Transverse	1139	1282	15.8	23.0		
LT					92.2	96.4
TL						95.7
Heat 6						
Longitudinal	1095	1264	19.3	33.5		
Long Transverse	1139	1264	18.9	35.2		
LT					85.3	100.7
TL						99.2

Table 6 Continued. Tensile and Toughness Test Results (Averages of Two Tests)
for Duplex Annealed Specimens of all Chemistries

Specimen Orientation	Tensile Tests				Fracture Toughness (MPa√m)	
	0.2 Percent Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation in 25.4 mm (Percent)	Reduction in Area (Percent)	K _{Id}	K _{Ic}
Heat 7						
Longitudinal	1104	1255	17.0	30.2		
Long Transverse	1104	1255	16.9	29.9		
LT						103.0
TL						100.8
Heat 8						
Longitudinal	1104	1255	15.5	30.3		
Long Transverse	1104	1255	15.9	27.2		
LT					88.3	97.0
TL						91.6

annealed Ti-6Al-6V-2Sn do not affect mechanical properties as strongly as would be the case for alpha-beta processed and recrystallized annealed Ti-6Al-6V-2Sn. The trend or direction of toughness variations with respect to chemistry differences agree generally with Kennedy's predictions, but the measured values indicate that the magnitude of change is smaller, especially for the yield strength. Also, there were no negative changes in fracture toughness, even with positive strength changes in tensile strength, which is contrary to the predictions of Kennedy. Nevertheless, it appears useful to have a quantitative relationship between fracture toughness and chemistry for material in the beta processed condition. Establishing such a relationship will require considerably more experimental effort.

More testing will probably be required to establish the statistical validity of the reported results for heats 2 through 8. However, the present results do indicate that a separate heat treat schedule will not be needed for each change in alloy chemistry to maintain strength and toughness within a specified range.

Mechanical Testing of Isothermal Quenched Material

Segments of the ring prolongations from heat 1 were given the isothermal quench heat treatment noted in Table 8. Each segment was subsequently machined into longitudinal tensile and LT orientation precracked Charpy impact specimens. The results of the tensile and impact tests are given in Table 8. The combinations of strength and toughness attained from the isothermal quench treatment are the same as those achieved from the solution treat and age treatments. This is best illustrated in Figure 16, where 0.2 percent yield strength is plotted against fracture toughness. The similarity of properties obtained from conventional solution treating and aging and from isothermal heat treating is also shown in Figures 13 and 14. These results show that no improved combinations of strength and toughness are obtained with these relatively complex isothermal quench techniques.

Fractography

Duplex Annealed Condition

Scanning electron microscopy showed that failure in duplex-annealed Ti-6Al-6V-2Sn was characterized by transgranular ductile, dimple formation (Figure 17). Further examination of specimens that had been nickel-plated, mounted, and polished so that the subsequent examination plane was perpendicular to the original fracture surface suggest that propagating cracks deflect locally at the boundaries of Widmanstätten alpha + beta colonies (Figure 18). The absence of any intergranular fracture is consistent with the findings of others, who showed that sub-transus annealing tends

Table 7. Predicted and Measured Variations in Strength and Toughness as a Function of Chemistry

Heat*	Predicted Change**		Measured Change	
	Yield Strength (MPa)	Fracture Toughness (MPa√m)	Yield Strength (MPa)	Fracture Toughness (MPa√m)
1	0	0	0	0
2	-267	+34	-36	+16.1
3	+27	-12	+9	+4.6
4	-98	+18	-27	+16.1
5	-9	+3	+18	+5.5
6	-80	+10	-18	+9.8
7	-80	+10	-9	+12.1
8	-80	+10	-9	+6.2

*Chemistry of each heat is defined in Table 1.

**From Reference 24.

to suppress the intergranular fracture mode that is normally observed in beta processed alpha-beta titanium alloys.²⁵ The absence of an intergranular fracture mode may be caused by the recrystallization of the transformed beta matrix which occurs during the sub-transus anneal. This phenomena has its parallel in the increase in fracture toughness observed in Ti-6Al-4V when the latter is given a recrystallization annealing treatment.

Solution Treated and Aged Condition

The decrease in tensile ductility and fracture toughness which accompanies increased strength in Ti-6Al-6V-2Sn can be associated with an increase in the amount of intergranular, flat fracture. These intergranular regions are generally observed only after aging at low temperatures (less than about 500°C). When viewed perpendicular to the principal fracture surface, cracks appear to follow the interface between the primary alpha particles along the prior beta grain boundaries and the beta matrix (Figure 19). This observation suggests that the interface phase, which has previously been shown to exist in the beta-extruded Ti-6Al-6V-2Sn alloy presently under investigation, may play an important, although as yet undetermined, role in controlling the fracture toughness of alpha-beta titanium alloys.

Table 8. Tensile and Toughness Test Results for Isothermal Annealed Specimens

	Longitudinal Tensile Specimens*		LT Fracture Toughness Specimens*	
	A	B	A	B
0.2 Percent Offset Yield Strength (MPa)	900	1096		
Ultimate Tensile Strength (MPa)	1020	1206		
Elongation in 25.4 mm (Percent)	14.8	7.2		
Reduction in Area (Percent)	24	15		
K_{Id} (MPa $\sqrt{m}$)			79.2	67.6

*Specimen A was solution treated 1 hr at 850°C and quenched directly to 600°C for 8 hr; specimen B was solution treated 1 hr at 850°C and quenched directly to 500°C for 8 hr. All specimens from heat 1 (Table 1).

The change in the appearance of the fracture surface as a function of aging temperature after solution treating at 900°C is shown in Figure 20. This figure shows the change from intergranular, flat fracture at the lower aging temperatures to transgranular ductile, dimple type fracture at the higher aging temperatures. Finally, it should be noted that the flat fracture regions observed in Ti-6Al-6V-2Sn which was aged at less than 500°C contain rather small equiaxed dimples. This can best be seen by comparing Figure 20a to 20b and Figure 20c to 20d.

Results

This investigation has shown that when beta-extruded Ti-6Al-6V-2Sn is given a subsequent sub-transus heat treatment, attractive combinations of strength and fracture toughness can result. The fracture toughness was found to be more sensitive to the sub-transus treatment than was the yield strength. Three essential levels of strength and toughness were achieved in beta-extruded Ti-6Al-6V-2Sn:

Figure 17. Scanning Electron Photomicrograph Taken Normal to Fracture Surface of Duplex Annealed Ti-6Al-6V-2Sn Tensile Test Specimen (Fracture is characterized by transgranular ductile fracture with large areas of dimple formation.)

At low strength levels (895 MPa yield stress), duplex annealed Ti-6Al-6V-2Sn has a fracture toughness of $82.5 \text{ MPa}\sqrt{\text{m}}$;

At intermediate strength levels (1034 MPa yield stress) Ti-6Al-6V-2Sn solution treated at 850°C and aged 4 hours at 600°C has a fracture toughness of $74.8 \text{ MPa}\sqrt{\text{m}}$; and

At high strength levels (1200 MPa yield stress) Ti-6Al-6V-2Sn solution treated at 900°C and aged 8 hours at 550°C has a fracture toughness of $54.9 \text{ MPa}\sqrt{\text{m}}$.

Figure 18. Optical Photomicrograph of Plane Normal to Fracture Plane of Duplex Annealed Ti-6Al-6V-2Sn Tensile Specimen (Crack growth occurs by local crack deflection at boundaries of Widmanstatten alpha-beta colonies.)

It appears that the fracture toughness increase associated with beta processing is caused by local crack rotation which occurs when the crack encounters the boundary between adjacent Widmanstatten alpha-beta colonies. Transmission electron microscopy has shown that, at the highest aging temperature examined, some ordering takes place in the primary alpha; however, this does not appear to be the principal strengthening mechanism in Ti-6Al-6V-2Sn. Precipitation of alpha in the previously transformed beta matrix appears to be more important. Also, some evidence is presented which suggests that as the strength level increases, the interface phase may play an increasingly important role in limiting the fracture toughness of alpha-beta titanium alloys.

Figure 19. Optical Photomicrograph of Plane Normal to Fracture Plane of Ti-6Al-6V-2Sn Tensile Specimen Solution Treated at 900°C, Water Quenched, Aged 8 Hours at 500°C, and Air Cooled (Note how the crack follows the interface between the primary alpha particles along the prior beta grain boundaries and the beta matrix.)

In addition, variations of alloy chemistry, within the limits allowed by MIL-T-9047, have been shown to have a negligible effect on tensile strength and only a slight effect on fracture toughness of duplex annealed material. This result is encouraging because it means that a separate heat treat schedule will not be required for such allowable deviations in alloy chemistry.

a. 900/400/8

100 μm

b. 900/400/8

13 μm

c. 900/450/4

100 μm

d. 900/450/4

13 μm

(NOTE AREAS OF FLAT FRACTURE CONTAIN SMALL AREAS OF EQUIAXED DIMPLES IN THESE FOUR PHOTOS)

Figure 20. Scanning Electron Photomicrographs of Fracture Surfaces at High and Low Magnification for Various Solution Treated and Aged Tensile Specimens

e. 900/500/8

100 μ m

f. 900/500/8

13 μ m

g. 900/550/8

100 μ m

h. 900/550/8

13 μ m

Figure 20 Continued. Scanning Electron Photomicrographs of Fracture Surfaces at High and Low Magnification for Various Solution Treated and Aged Tensile Specimens

Figure 20 Continued. Scanning Electron Photomicrographs of Fracture Surfaces at High and Low Magnification for Various Solution Treated and Aged Tensile Specimens

ACCOMPLISHMENTS

The strength and toughness relationships of duplex annealed and solution treated and aged beta-extruded Ti-6Al-6V-2Sn were established for eight variations of chemistry within the limits allowed by MIL-T-9047 as a result of this program. Baseline mechanical properties for the planned application were determined. A preliminary understanding of the relationship between mechanical properties and microstructure also has resulted from this study. Several areas requiring further investigation, such as a statistical analysis of the effects of alloy chemistry variations on heat treat response, the relationship of K_{Ic} to K_{Id} , the effects of interface alpha on properties, as well as others, have been identified. Bendix is currently investigating some of these areas.

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