Basic Properties of Thin-Film Diffusion-Brazed Joints in Ti-6A1-4V

Thin-film diffusion brazing of Ti-6A1-AV with copper offers attractive potentials for producing high-integrity joints under reasonable manufacturing conditions

BY ALLAN H. FREEDMAN

ABSTRACT. The objective of this investigation was to determine some of the basic metallurgical and mechanical properties of Ti-6A1-4V joints produced by thin-film diffusion-brazing with copper. Critical properties which were determined included tensile, fatigue, fracture toughness, stress corrosion, and corrosion-fatigue behavior of butt joints, shear strength of lap joints, and comparative properties of Ti-6A1-4V base metal. In addition, fracture modes of joints and base metal were determined from specimens exposed to tensile overload, fatigue, stress corrosion, and corrosion fatigue.

The results of this program indicate that most of the properties of thin-film diffusion-brazed joints equal or exceed base metal properties. One property, plane-strain fracture toughness (K_{Ic}) is lower for joints but the threshold stress-intensity factor for stress corrosion (K_{Iscc}) is somewhat higher for joints than for base metal. In addition, the thin-film diffusion-brazing process offers much better versatility and substantial manufacturing advantages over solid-state diffusion bonding. Thus, the process offers very attractive potential for producing high-integrity joints under reasonable manufacturing conditions, and it should find wide use for joining titanium-alloy structures.

Introduction

A thin-film diffusion-brazing process has been developed for joining titanium-alloy components ranging from thin-gage honeycomb structures to heavy plate-type configurations.[1] The process depends upon the placement of a thin film (8 to 50 mg/in.^{2}) of copper foil or electroplated copper in the joint area. The joint is heated to 1700° F under modest pressure to assure contact of the surfaces to be joined. During the joining cycle, diffusion between the titanium and copper takes place rapidly to form a copper-titanium eutectic composition, which melts at 1635° F and wets the titanium. Like a conventional brazing filler metal, the eutectic liquid fills the space between faying surfaces. Continued diffusion between the liquid and base metal occurs rapidly, changing the composition of the liquid, and thereby increasing its melting temperature so that it soon solidifies at the 1700° F brazing temperature. Further holding at 1700° F or at some lower temperature permits additional diffusion to occur until the maximum residual copper content of the joint is reduced sufficiently to assure good mechanical properties.

The amount of copper required to achieve a good joint varies somewhat, depending upon the geometry of the joint, joint fit-up tolerances, pressure applied to the joint, and heating rate. For butt or lap joints, the amount of copper needed is dependent primarily upon the void volume in the joint which must be filled with liquid. Thus, flatness, surface finish, and applied pressure will affect the void volume and the amount of copper required. The amount of copper needed for a tee joint is dependent upon fit-up tolerances as well as the amount of liquid required to form a fillet of the desired size.

A secondary factor affecting the amount of copper needed to produce a joint is the heating rate to the bonding temperature. Higher heating rates will decrease the amount of solid-state copper diffusion into the substrate, thereby leaving more of the copper available for formation of eutectic liquid. Thus, as heating rates are increased, the amount of copper needed to obtain a good bond will decrease. Once the amount of copper needed to produce a joint is established, it is necessary to establish the diffusion time at the bonding temperature which will reduce the maximum residual-copper content of the joint to the desired level.

In most cases, copper levels of 8 to 50 mg/in.,^{2} heating rates of 400° F/hr to 4,000,000° F/hr, pressures of 10 to 100 psi, and bonding times of 1 to 12 hr at 1700° F cover the range of parameters which can be used to produce most joint geometries. Regardless of the parameters used to produce joints, they should exhibit similar properties if they contain the same residual-copper content after fabrication.

The objective of this program was to determine some of the basic metallurgical and mechanical properties of thin-film diffusion-brazed joints in Ti-6A1-4V. Critical properties which were determined include tensile, fatigue, fracture toughness, stress corrosion, and corrosion-fatigue behavior of butt joints, shear strength of lap joints, and comparative properties of Ti-6A1-4V base metal. In addition, fracture modes of joints and base metal were determined from specimens exposed to tensile overload, fatigue, stress corrosion, and corrosion fatigue.

Materials and Procedures

Materials

Annealed Ti-6A1-4V plate of 1.03 to 1.07 in. thickness was used to fabricate most of the specimens used in this work. The material came from heat A41641-1 produced by G.O. Carlson Company. A few tests were conducted to compare the effects of the joining cycle on tensile properties of sheet and plate. The annealed Ti-6A1-4V sheet material used for these

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Table 1—Vendor-Supplied Chemical Analyses of Ti-6Al-4V Sheet and Plate, %

Element	1 in. plate[a]	0.062 in. sheet[b]
Al	6.41	6.0
V	4.26	4.0
C	0.021	0.023
O	0.14	0.11
N	0.018	0.012
H	0.0009	0.0055
Fe	0.13	0.06

[a] G. O. Carlson Co., Heat A41641-1.
[b] Titanium Metals Corp. of America, Heat 5375.

tests and for lap shear tests was 0.062 in. thick, and it came from heat 5375 produced by Titanium Metals Corporation of America. Chemical analyses obtained from the suppliers of the materials are shown in Table 1.

These materials or specimens prepared from these materials were chemically cleaned before exposing them to a joining cycle at elevated temperatures. Cleaning consisted of :

1. Solvent degreasing and/or scrubbing with a commercial cleanser.
2. Hot alkaline cleaning.
3. Tap-water rinsing.
4. Immersion for 1-3 min in 2-5HF, 20-23HNO_3, balance H_2O at room temperature.
5. Tap-water rinsing.
6. Distilled water rinsing.
7. Drying in still air.

In most cases, electrolytic tough-pitch copper foil weighing 50 mg/in.^{2} was used for joining. Prior to joining, the foil was cleaned by:

1. Degreasing in methyl ethyl ketone.
2. Immersing for 1 to 3 min in 25 oz $CuSO_4$–8 oz H_2SO_4–1 gallon H_2O (copper plating solution) at room temperature.
3. Tap-water rinsing.
4. Distilled water rinsing.
5. Drying in still air.

In some cases, this same solution was used to apply copper to the Ti-6Al-4V by electroplating.

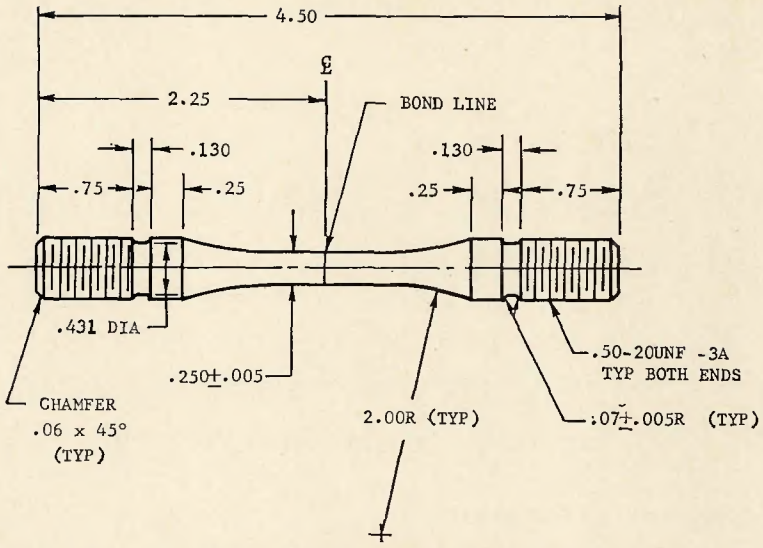


Fig. 1—Butt-joint fatigue specimen. Notes: (1) all diameters concentric within .001; (2) finish machine with a sharp tool and light cuts to prevent bending, overheating, and cold working; (3) polish reduced section in stages with emery paper from No. 0 to No. 000 (do not buff)

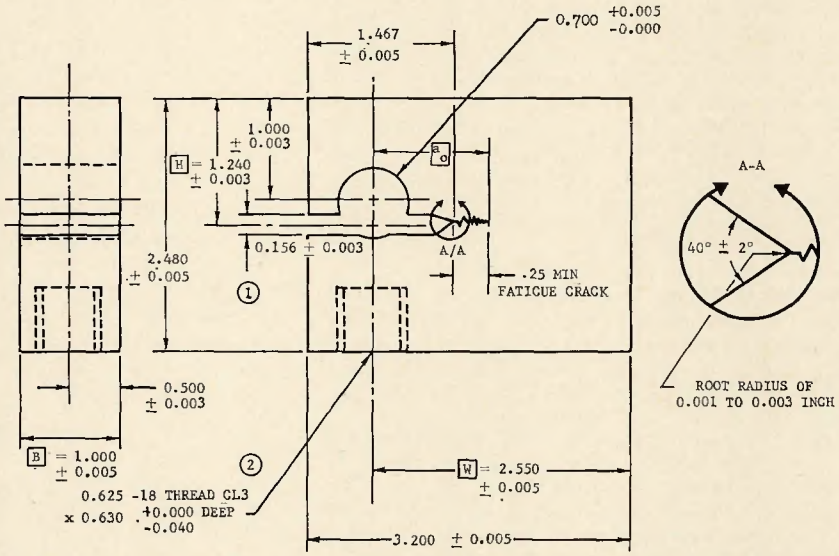


Fig. 2—1T-type wedge-opening-loading (WOL) specimen for base metal. Notes: (1) bore and slot perpendicular to threads and face within 0.003; (2) center line of thread must coincide with center line of bore within 0.003; (3) both 3.200 × 2.480 surfaces to be 32| or better with machine; (4) markings perpendicular to fatigue-crack axis; (5) do not grind any surface

Specimen Fabrication

Rectangular specimens approximately 2 in. long by 1 in. wide by $^1/_2$–$^3/_4$ in. high were used to determine the relationships between bonding pressure and copper content required to produce a complete bond over a large faying surface. The 1 by 2 in. faying surfaces were machined flat and parallel within ±0.0005 in. by flycutting rather than grinding because Carter[2] has reported that grinding of Ti-6Al-4V produces damage which results in a marked reduction in fatigue strength. Whether this damage would be retained on surfaces joined by brazing was not known, and flycutting was employed to avoid the possibility of this problem. The specimens were joined in a cold-wall vacuum furnace equipped with a compression cage to permit loading of the specimens by hanging weights on an external loading rod.

To produce butt-joint specimens for tensile, fatigue, and smooth-specimen stress-corrosion tests, blocks approximately 5 in. wide by 2$^1/_2$ in. high by 1 in. thick were bonded along the 5 by 1 in. surfaces to yield a blank 5 in. high by 5 in. wide by 1 in. thick. The faying surfaces of these blocks were machined flat and parallel within ±0.0005 inch by flycutting prior to bonding. Copper foil (50 mg/in.^{2}) was placed between the blocks, and bonding was accomplished by vacuum induction heating in a hot press. After bonding, the blanks were sectioned to provide tensile and stress corrosion specimens with a $^1/_4$ in. diameter reduced section conforming to a Type R3 tensile specimen specified by Federal Test Method No. 151a. Specimens for fatigue testing were machined to the dimensions shown in Fig. 1. In all cases, the specimens were oriented with the rolling direction of the base metal parallel to the loading direction.

Fracture toughness, fatigue, and stress corrosion tests on base metal were performed on wedge-opening-

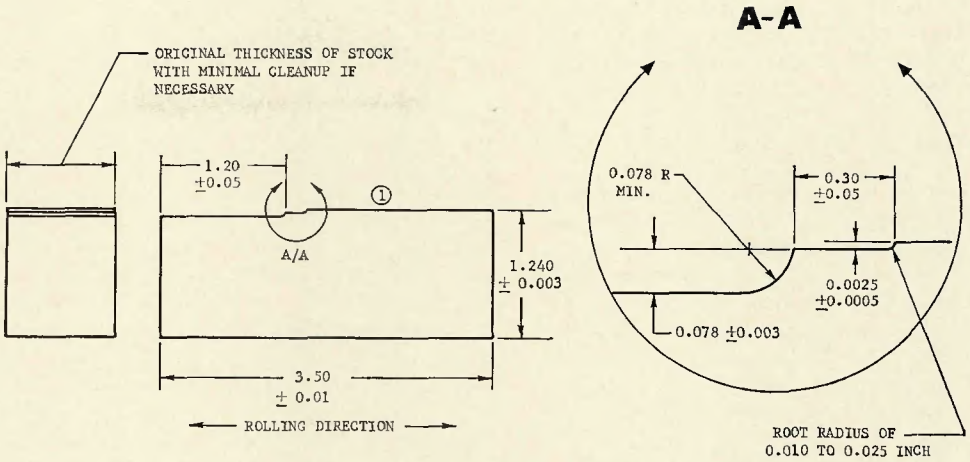


Fig. 3—Blocks for fabricating WOL—specimen joints. Notes: (1) finish after milling slots—fly cut surface to ± 0.0005 flatness; (2) blocks to be machined in matched pairs; (3) do not grind any surface

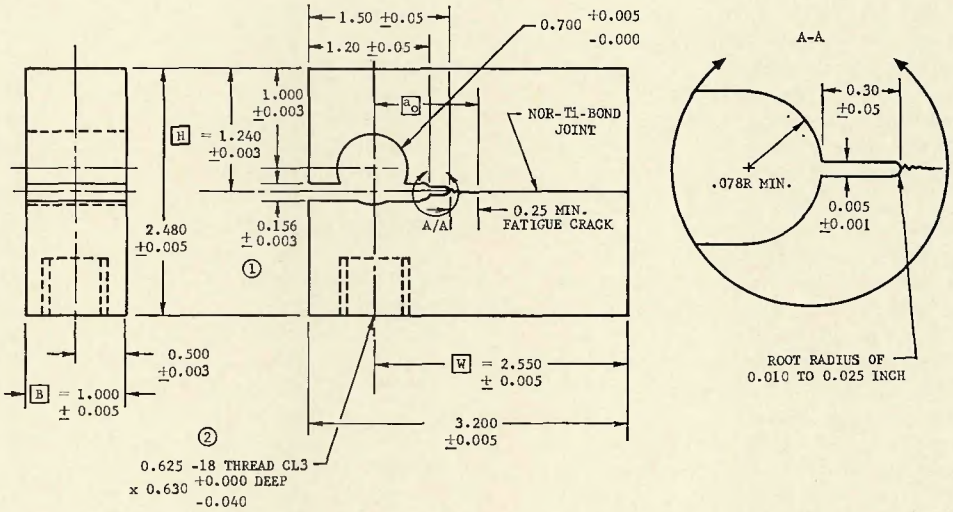


Fig. 4—1T-type wedge-opening-loading (WOL) specimen for joints. Notes: (1) bore and slot perpendicular to threads and face within 0.003; (2) center line of thread must coincide with center line of bore within 0.003; (3) both 3.200 × 2.480 surfaces to be 32| or better with machine markings perpendicular to fatigue-crack axis; (4) do not grind any surface

loading (WOL) specimens of the 1T-type shown in Fig. 2. Specimens were tested with the loading direction parallel to the long-transverse direction with the crack growing in the rolling direction (WR Orientation) as well as the opposite case where the load was parallel to the rolling direction with the crack growing in the long-transverse direction (RW Orientation). Specimens were tested in the annealed condition and after a thermal exposure simulating the joining cycle; the thermal exposure was performed in a cold-wall vacuum furnace.

Similar specimens were used to measure the same properties of joints. Blocks were premachined to the configuration shown in Fig. 3 with the base metal in the WR Orientation. Pairs of blocks were then joined in the vacuum hot press using the copper foil of 50 mg/in.^{2} placed across the faying plane and the same bonding approach used for butt-joint specimens. After bonding, the specimens were finish machined to the dimensions shown in Fig. 4. These dimensions were similar to the dimensions of base metal specimens; the exception was the notch geometry which was necessarily modified to assure that the fatigue crack would initiate and grow in the joint. The length of the premachined blanks was approximately 0.3 in. larger than the finish length. This excess material was cut from each specimen to provide a section for metallographic examination and electron-microprobe analyses to assess bond quality.

Unless noted otherwise, all of the butt joints were produced using one layer of the 50 mg/in.^{2} copper foil placed across the faying surfaces. The specimens were loaded to 100 psi compressive stress at the start of the bonding cycle unless noted otherwise. The standard bonding cycle adapted for this work consisted of heating at any convenient rate to 1300° F, heating at 400° F/hr to 1700° F, holding 4 hr at 1700° F, and cooling from 1700° F to 1300° F in 90 minutes and cooling from 1300 to 1100° F in 60 minutes.

Lap-shear specimens were produced by joining one large lap-shear panel which was cut to yield specimens 8 in. long and 3/4 in. wide with a one-T (0.062 in.) overlap. The rolling direction of the base metal was oriented parallel to the loading direction. One double of 0.062 in. sheet, $1^3/_8$ in. long by $^3/_4$ in. wide was located at each end of each specimen on opposite sides to reduce eccentricity during loading. The lap area and the doublers were bonded simultaneously using 30 mg/in.^{2} of electroplated copper in the faying surfaces. Bonding was accomplished in an evacuated and sealed envelope of stainless-steel foil; this was placed in small ceramic tools and heated at 500° F/hr, held 1 hr at 1700° F, and cooled at the same rate used for butt joints. The effective bonding pressure on the faying surfaces was 16 psi.

A few tests were conducted to compare tensile properties of 1 in. thick plate and 0.062 in. thick sheet in the as-received (annealed) condition and after exposure to the standard thermal cycle used for bonding. The thermal cycle was performed in a cold-wall vacuum furnace. Tensile specimens were 0.505 in. diameter round bars and 0.500 in. wide flat specimens for the plate and sheet, respectively. These specimens conformed to the Type R1 bar and Type F2 sheet specimens specified by Federal Test Method Standard No. 151a.

Specimen Testing

All of the base metal and butt-joint

tensile tests and fracture-toughness tests on WOL specimens were conducted using a screw-type tensile machine operating at a head speed of 0.05 ipm. A compliance gage was used to detect "pop-in" during the fracture-toughness tests.

Lap-shear tests were conducted in a hydraulic tensile machine equipped with wedge-action grips.

Fatigue tests on butt-joint and base metal specimens were performed in an electro-hydraulic fatigue machine operating at 1200 cpm under sinusoidal tension-tension loading at a stress ratio of 0.1. Cracking of WOL specimens for fracture-toughness tests was also performed in the same machine using the same parameters; an exception was that the minimum load was held at 200 lb, and the maximum load was varied from 4000 to 7000 lb as required to produce a fatigue crack of the desired length within approximately 20,000 cycles.

Fatigue-crack growth studies on WOL specimens tested in air were also performed using the same fatigue machine. The specimens were precracked using the same parameters used for the fracture-toughness tests. The stress ratio was then changed to 0.1 and fatigue-crack length vs. number of cycles was measured. Grid lines were scribed on these specimens at a spacing of 0.05 in. and optical measurements were used to determine the number of cycles required for each 0.05 in. increment of crack growth.

The same basic approach was used to study fatigue-crack growth in 3.5% NaCl (corrosion fatigue). A transparent plastic container surrounded the specimen and contained the corrodent. In addition, the loading clevis and pin, which were partially immersed in the corrodent, were made of annealed Ti-6Al-4V to minimize the possibility of galvanic effects between the specimen and loading fixtures. The specimen was precracked in salt water before taking measurements of crack-growth behavior.

Stress corrosion tests in 3.5% NaCl were conducted on WOL specimens using the same fixtures used for the corrosion-fatigue tests. Each specimen was precracked in 3.5% NaCl and then transferred to a dead-weight level-arm loading fixture where it was loaded below K_{ISCC}, the threshold stress-intensity factor for stress corrosion. The load was increased at periods varying from 1 to 72 hr if optical measurements showed no evidence of crack growth. Once a load was reached at which crack growth occurred by stress corrosion, the load was held constant until the specimen failed.

Stress corrosion tests on smooth butt-joints were conducted using the same type of specimens used for the tensile and fatigue tests. These tests were conducted at constant load using the same dead-weight loading fixture used for the WOL specimens.

Results and Discussion

Bonding Parameters for Large Faying Surfaces

Before fabricating the butt-joints and WOL (wedge-opening-loading) specimens required for this program, it was necessary to establish the bonding pressure and copper quantity needed to produce complete bonds across large faying surfaces between thick sections. The results of these studies are summarized in Table 2. A pressure greater than 25 psi, but no greater than 50 psi, was required to obtain a complete bond when using one layer of copper foil weighing nominally 50 mg/in.^{2} The specimens bonded with 49 to 61 mg/in.^{2} of electroplated copper required more than 50 psi, but no more than 100 psi, to produce a complete bond. The reason for the higher pressure needed with electroplated copper is not known, but may be related to the more uniform thickness of the copper foil compared to the electroplated copper.

It should be appreciated that these bonding parameters apply to flycut surfaces, flat and parallel within ±0.0005 in., and with a surface finish of approximately 32 rms. Changes in surface finish and surface flatness would be expected to affect the relationships between bonding pressure, copper quantity, and degree of bonding. In addition, heating rates differing from the 400° F/hr used in these studies should affect the amount of copper needed to form a good bond. Higher heating rates will decrease the amount of solid-state copper diffusion into the substrate, thereby leaving more of the copper available for formation of eutectic liquid. Thus, as heating rates are increased, the amount of copper needed to obtain a good bond will decrease.

It appears that a good bond could be obtained with less copper if the bonding pressure were increased, although this effect was not studied for two reasons. First, low bonding pressures are a distinct manufacturing advantage and, second, pressures above 100 psi would produce significant dimensional changes from creep at the 1700° F bonding temperature. Based upon these considerations, the butt joints used in this work were produced using one layer of copper foil weighing 50 mg/in.^{2} and a bonding pressure of 100 psi unless otherwise noted.

It should be noted that with these bonding parameters, there was no excess liquid squeezed out along the periphery of the joints. In fact, the joints ended approximately 0.003 in. from the edges. This effect probably was caused by movement of liquid from the edges inward to fill voids. A finish-milling operation was performed to remove 0.010 to 0.015 in. from the surfaces to provide specimens which contained complete joints across the faying surfaces. The use of approximately 5% more copper probably would have permitted formation of sufficient eutectic liquid to produce joints to the edges of the faying surfaces.

A few preliminary experiments were conducted to bond large faying surfaces of 0.062 in. sheet with no machining of the sheet surfaces. Using the same bonding parameters used for plate, excess eutectic liquid squeezed out of the joint. Thus, the bonding pressure and copper quantity used for heavy plate were greater than needed for sheet material. This effect is believed to be caused by the ability of sheet material to bend and conform, permitting closer contact along the faying surface than obtainable with rigid material.

Table 2—Bonding Parameters for Large Faying Surfaces Between 1 in. Thick Plates

Specimen no.	Bonding pressure,[a] psi	Total copper quantity, mg/in.^{2}	Notes
Copper foil			
2	15	38	60% bonded
4	100	51	Excellent bond
W2	75	52	Excellent bond
W3	60	52	Excellent bond
5	50	54	Excellent bond
1A	50	50	Excellent bond
2A	50	50	Excellent bond
6	25	52	Outer edges unbonded
Electroplated copper			
8	100	61	Excellent bond
7	50	59	Outer edges unbonded

[a] All specimens heated at 400° F/hr from 1300–1700° F.

Fracture Toughness

Data on plane-strain fracture toughness are summarized in Table 3. Fracture toughness of thin-film diffusion-brazed joints containing 50 mg Cu/in.^{2} was measured after different fabrication cycles to determine the relationships between processing parameters and joint toughness. All of the specimens were made with the base metal in the WR Orientation in which the load was perpendicular to the rolling direction and the crack grew parallel to the rolling direction.

Specimen W-8, held a total of 1 hr at 1700° F, contained a maximum residual copper content of 9.5% at the center of the joint, based upon electron-microprobe analysis. Fracture toughness of the joint was compratively low. Specimen W-11, held 4 hr at 1700° F, contained a maximum residual copper content of 7.1% while specimens W-6 and W-26, held 12 hr at 1700° F, contained 5.9 and 5.4% copper, respectively. Since specimens W-9, W-10, and W-12 were also bonded for 4 hr at 1700° F, it is probable that residual copper levels in the joints were approximately 7%. The influence of residual-copper content on fracture toughness is obvious. These results permit us to establish quantitatively the relationships between time at 1700° F, maximum residual-copper content, and fracture toughness of the joints—Fig. 5.

The fracture toughness of specimens W-11 and W-12 was somewhat higher than the toughness of specimens W-9 and W-10, even though both sets of specimens were held a total of 4 hr at 1700° F. However, the bonding pressure was maintained during the entire cycle for specimens W-11 and W-12, whereas specimens W-9 and W-10 received an interrupted bonding cycle and the pressure was only maintained during the first 30 minutes at 1700° F. Specimens W-19 and W-20 were given a four-hour cycle at 1700° F, with the pressure removed after the first 30 minutes, to determine if the interruption in the bonding cycle or the time under pressure at 1700° F was responsible for these differences in toughness. Since the toughness values for W-19 and W-20 fell between the values for W-9 and W-10 and W-11 and W-12, it was difficult to make definite conclusions considering the limited number of tests. Nevertheless, it appears that maintaining pressure during the entire bonding cycle provides slightly improved fracture toughness. This effect may be related to a small amount of deformation produced by creep.

Specimens W-21 and W-27 were tested to determine the influence of cooling rate from the bonding temperature on fracture toughness. The specimens were cooled from 1700 to 1000° F in 15 minutes. The fracture toughness produced by this cooling rate was approximately 10% less than the fracture toughness of specimens W-11 and W-12, which employed the same bonding conditions except for the slower controlled cool. These tests indicate that fracture toughness of the joints does not appear to be highly sensitive to cooling rate.

Based upon these fracture-toughness data, the balance of this work was standardized on a diffusion treatment of 4 hr at 1700° F under 100 psi compressive stress followed by the controlled cool to provide a good combination of manufacturing practicality and reasonable fracture-toughness. This cycle was based upon the use of 50 mg Cu/in.^{2} on faying surfaces. If less copper can be used successfully in manufacturing operations, then the time at 1700° F could be reduced to achieve the same mechanical properties.

Fracture-toughness tests on as-received (annealed) base metal (Table 3) showed similar values for the WR and RW Orientations. The results on specimens exposed to the fabrication cycle clearly show an increase in toughness. This increase was large enough that the specimen thickness of 1 in. no longer met the ASTM recommendation for ensuring plane-strain conditions. Thus, the values found are somewhat higher than would result from a thicker specimen, and they are not true K_{IC} values. The fracture toughness of the joints produced using the chosen processing cycle (4 hr at 1700° F) was approximately 60% of the fracture toughness of the as-received material. However, by varying the maximum residual-copper content between 5.4 and 9.5% in the joints, the fracture toughness may range from approximately 40% to 90% of the fracture toughness of as-received (annealed) Ti-6Al-4V.

The longitudinal microstructure of the as-received (annealed) Ti-6Al-4V plate used to fabricate the WOL spec-

Table 3—Fracture Toughness of Base Metal and Joints

Specimen no.	K_{IC},[a] (ksi√in.)	K_{IX},[b] (ksi√in.)	Max. residual Cu in joint, wt-%	Fabrication cycle and notes[c]
Joints (WR orientation of base metal)				
W-8	29.8	31.5	9.5	30 minutes at 1700° F under 100 psi compression, furnace cool, +30 minutes at 1700° F under no load, controlled cool.
W-9	40.2	42.9	—	30 minutes at 1700° F under 100 psi compression, furnace cool, +3½ hr at 1700° F under no load, controlled cool.
W-10	39.5	39.5	—	
Average	39.8	41.2		
W-11	44.7	50.2	7.1	4 hr at 1700° F under 100 psi compression, controlled cool.
W-12	45.9	54.7	7.1 (estimated)	
Average	45.3	52.5		
W-19	41.6	43.0	—	30 minutes at 1700° F under 100 psi compression +3½ hr at 1700° F under no load, controlled cool.
W-20	43.8	47.4	—	
Average	42.7	45.2		
W-6	68.8	71.5	5.9	12 hr at 1700° F under 100 psi, controlled cool.
W-26	60.1	61.2	5.4	
Average	64.5	66.4		
W-21	40.0	41.3		4 hr at 1700° F under 100 psi compression, cooled from 1700° to 1000° F in 15 minutes.
W-27	42.0	47.0		
Average	41.0	44.2		
Base metal				
TWR-1	75.4	81.9		As-received (annealed)—specimens oriented with load parallel to long transverse direction and crack growing in rolling direction (WR orientation).
TWR-2	80.2	84.5		
Average	77.8	83.2		
TRW-1	68.6	72.6		As-received (annealed) — specimens oriented with load parallel to rolling direction and crack growing in long-transverse direction (RW orientation).
TRW-2	78.1	83.6		
TRW-3	77.7	80.6		
Average	74.8	78.9		
TWR-7[d]	97.0	108		Exposed to fabrication cycle (4 hr at 1700° F, controlled cool). Same orientation as TWR-1, -2.
TWR-8[d]	85.0	109		
Average	91.0	109		
TRW-4[d]	91.5	116		Exposed to fabrication cycle (4 hr at 1700° F controlled cool). Same orientation TRW-1, -2, -3.
TRW-5[d]	98.8	118		
Average	95.7	117		

[a] Plane-strain fracture toughness.
[b] Plane-strain stress-intensity factor based upon maximum load.
[c] All specimens heated in 400° F/hr from 1300–1700° F. Controlled cool is 1700–1300° F in 90 minutes and 1300–1100° F in 60 minutes.
[d] K_{IC} values are invalid (somewhat high) because the specimen thickness was insufficient to meet ASTM recommendations.

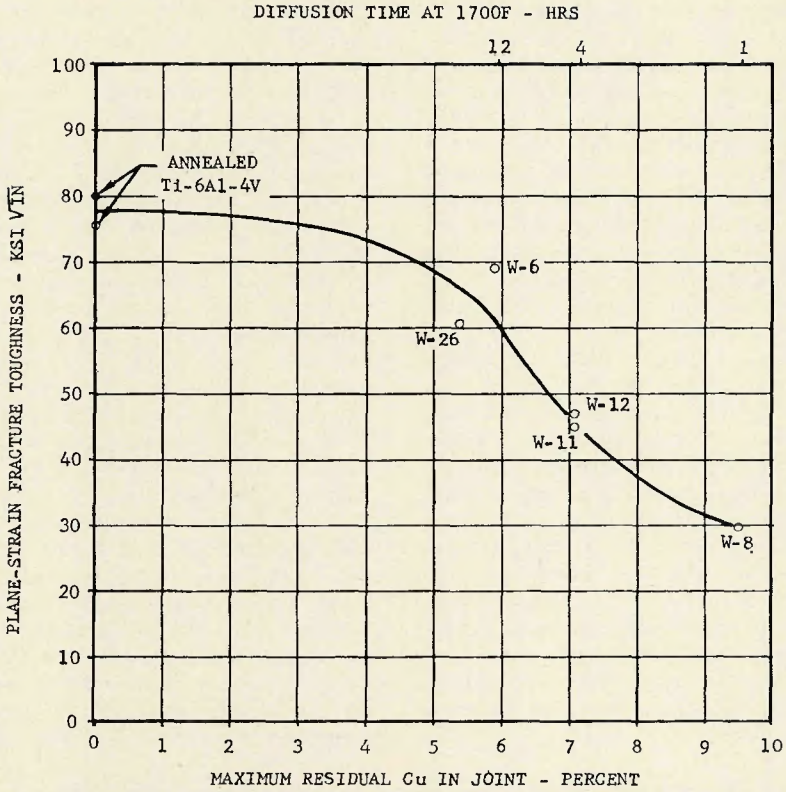


Fig. 5—Effect of residual copper content and diffusion time at 1700° F (for 50 mg/in.^{2} Cu) on fracture toughness of joints

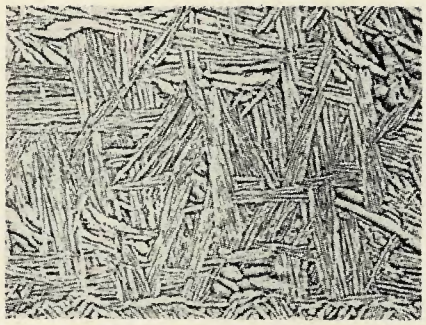
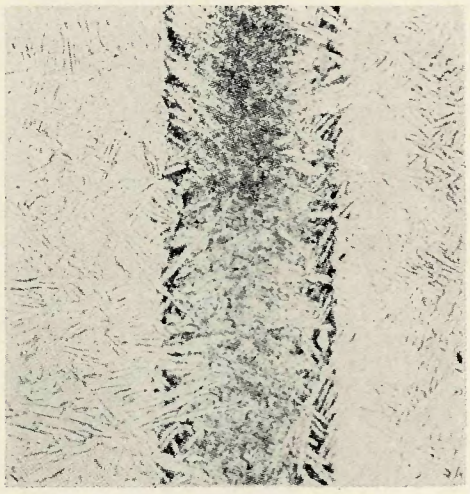


Fig. 6—Microstructure of as-received (annealed) Ti-6Al-4V plate material. X200

imens is shown in Fig. 6. The structure was predominantly transformed beta, indicating that a comparatively high annealing temperature was used by the producer. In addition, the structure also showed evidence of alloy segregation.

Typical microstructures and microhardness values for joints bonded 1 and 4 hr at 1700° F followed by controlled cooling are shown in Fig. 7. The joints consist of the proeutectoid alpha and eutectoid phases, with the eutectoid composed of alpha + Ti_2Cu. The joint bonded 4 hr at 1700° F to yield 7.1% residual copper at the center contained approximately 60% alpha and 40% eutectoid in this region. The joint bonded 1 hr at 1700° F to yield 9.5% residual copper at the center contained approximately 5% proeutectoid alpha and 95% eutectoid. Phase diagram data[3] show that the eutectoid composition in the Ti-Cu system occurs at 7.1 wt-% copper. The microstructures of the joints indicate that the eutectoid composition in the Ti-6Al-4V/Cu system is apparently greater than 7.1% copper and may be as high as 10%. Both joints exhibited similar maximum microhardness values, although the center of the joint bonded 1 hr at 1700° F was somewhat softer.



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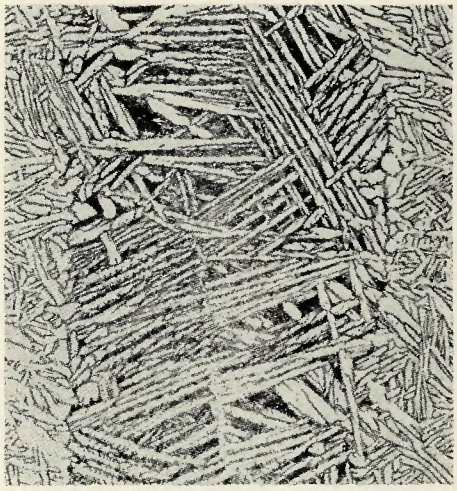
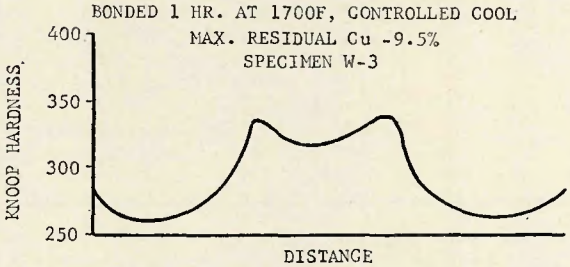




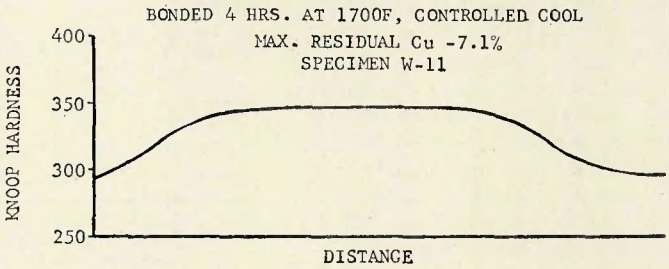
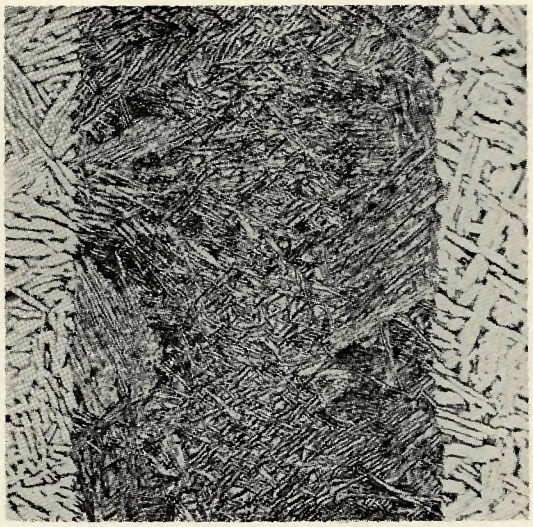


Fig. 7—Microstructures and microhardnesses of butt joints bonded 1 and 4 hr at 1700° F followed by controlled cool reduced 30% on reproduction)



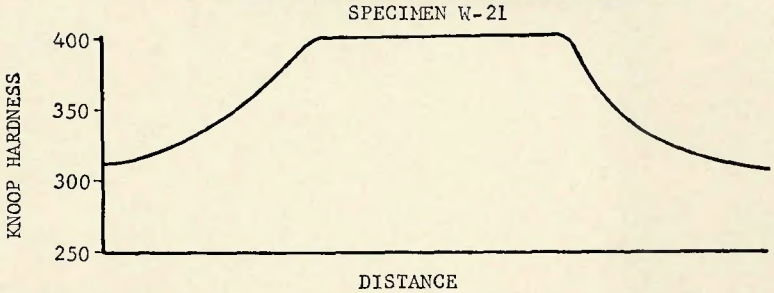


Fig. 8—Microstructure and microhardness of butt joint bonded 4 hr at 1700° F and cooled to 1000° F in 15 minutes. X200 (reduced 26% on reproduction)

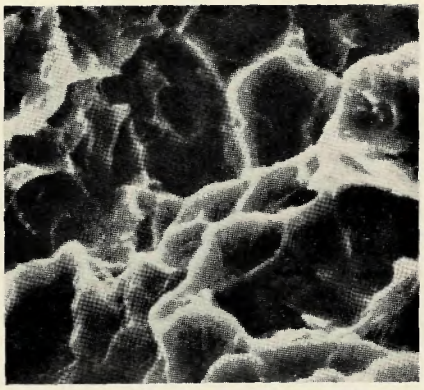


Fig. 9—Fracture surface of annealed Ti-6Al-4V tested in WR orientation—specimen TWR-1. X3000 (reduced 42% on reproduction)

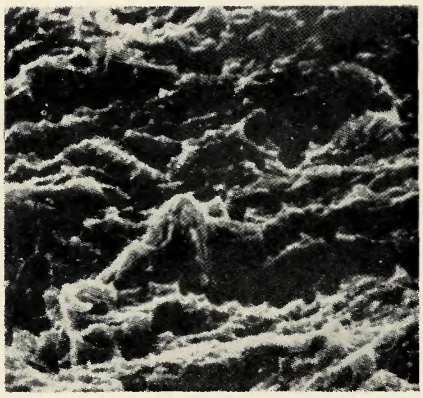


Fig. 10—Fracture surface of joint containing 9.5% maximum residual copper (bonded 1 hr at 1700° F)—specimen W-3. X3000 (reduced 42% on reproduction)

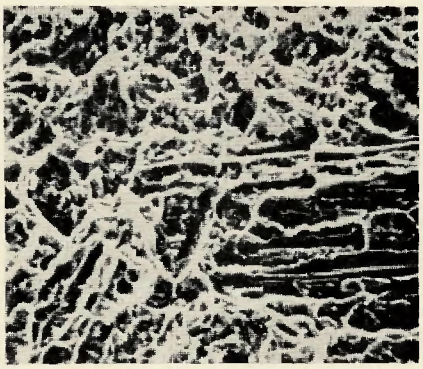


Fig. 11—Fracture surface of joint containing 7.1% maximum residual copper (bonded 4 hr at 1700° F)—specimen W-12. X1000 (reduced 33% on reproduction)

Figure 7 shows that the proeutectoid alpha in the joint containing 7.1% residual copper is much coarser than in the joint containing 9.5% residual copper. This probably results from the lower alpha + beta-to-alpha transformation temperature in the joint containing higher copper. Since the transformation of beta to alpha plus beta occurs by a nucleation and growth process, the lower transformation temperature should favor finer transformation products.

Figure 8 shows the microstructure and microhardness values for a joint bonded 4 hr at 1700° F and furnace cooled from 1700 to 1000° F in 15 minutes. While the primary beta grains were coarse, the alpha needles were very fine compared to the structure produced by the controlled cool. Even though the joint hardness was somewhat higher, it was surprising that this fine structure did not exhibit better toughness than the coarser structure produced by the standard controlled cool. It appears that the toughness of the joint is more sensitive to copper content than to grain size.

Several of the fracture-toughness specimens were examined by scanning electron fractography to permit comparison of fracture modes for base metal and joints. The fracture surface of annealed base metal tested in the WR Orientation (Specimen TWR-1) shows a typical transgranular dimple overload structure—Fig. 9. The dimples are large with some evidence of shearing, which is to be expected from the loading mode for a WOL specimen.

Figure 10 shows the fracture mode of a joint processed in the same manner as specimen W-8 (Table 3) to yield a maximum residual-copper content of 9.5% (1 hr at 1700° F, controlled cool). The fracture surface shows a transgranular dimple overload rupture, but the dimples are very small and flat, indicating significantly lower ductility than the annealed base metal. The dimples are too fine to detect any evidence of shear.

Fractures from specimens W-9 and W-12, which received different processing (Table 3) but the same total time (4 hr) at 1700° F to yield a maximum residual-copper level of 7.1% were examined. No significant differences were observed between the fracture modes of these two specimens. Figure 11 shows the fracture appearance for specimen W12. The fracture is inhomogeneous because of differences in microstructure across the joint. Areas of transgranular dimple rupture are present along with fracture at the interface between the alpha needles and the matrix. The surface of specimen W-21 (microstructure shown in Fig. 8), which was cooled more rapidly, also exhibited an inhomogeneous fracture (Fig. 12). Approximately two-thirds of the surface showed a transgranular, ductile, dimple rupture with dimples which appeared more ductile than those exhibited by specimen W-12, which was control cooled. However, the remaining third of the surface showed a flat, transgranular structure that appeared to be cleavage in nature.

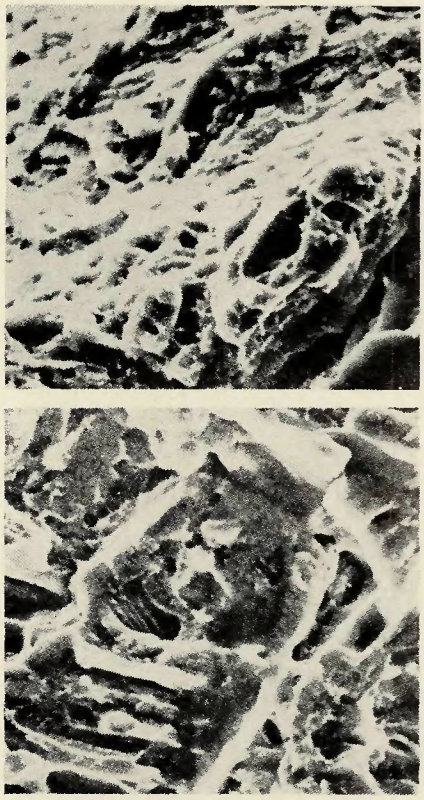


Fig. 12—Fracture surface of joint bonded 4 hr at 1700° F and cooled to 1000° F in 15 minutes. A (top)—transgranular, ductile dimple rupture; B (bottom)—flat, transgranular cleavage-like rupture. X2000 (reduced 38% on reproduction)

Table 4—Tensile Properties of Ti-6Al-4V Sheet and Plate

Specimen No.	Tensile strength, psi	0.2% yield strength, psi	Elongation, %	Reduction of area %	Modulus, psi	Notes
1 in. plate—as-received (annealed)						
TR-1	143,300	134,600	13.0	24.9	17.3×10^6	Longitudinal
TR-2	141,800	134,800	13.0	23.1	17.2×10^6	orientation
Average	142,600	134,700	13.0	24.0	17.3×10^6	
TW-1	143,000	135,100	15.0	24.6	17.1×10^6	Transverse
TW-2	142,700	134,300	15.0	25.4	18.2×10^6	orientation
Average	142,900	134,700	15.0	25.0	17.7×10^6	
1 in. plate—exposed to fabrication cycle[a]						
TR-3	138,100	124,800	17.0	31.6	17.2×10^6	Longitudinal
TR-4	135,500	121,400	17.0	35.9	17.1×10^6	orientation
Average	136,800	123,100	17.0	33.8	17.2×10^6	
TW-3	138,800	124,800	17.0	40.1	17.4×10^6	Transverse
TW-4	138,700	124,300	17.0	38.4	17.1×10^6	orientation
TW-5	139,400	125,800	17.5	37.5	17.4×10^6	
Average	139,000	125,000	17.2	38.7	17.3×10^6	
0.062 in. sheet—as received (annealed)						
1	140,400	137,100	15.5	—	16.3×10^6	Longitudinal
2	140,900	137,900	15.5	—	16.4×10^6	orientation
Average	140,700	137,500	15.5		16.4×10^6	
0.062 in. sheet—exposed to fabrication cycle[a]						
3	131,900	125,300	17.0	—	16.7×10^6	
4	132,500	125,400	18.5	—	16.8×10^6	Longitudinal
5	132,300	125,600	17.0	—	16.9×10^6	orientation
6	132,100	125,000	17.5	—	16.7×10^6	
Average	132,200	125,300	17.5		16.8×10^6	
0.062 in. sheet—exposed to modified fabrication cycle[b]						
A1	134,500	120,100	16.0	—	18.3×10^6	
A2	135,500	128,000	15.0	—	18.3×10^6	Longitudinal
A3	134,500	126,400	15.5	—	17.9×10^6	orientation
A4	133,600	126,100	16.5	—	18.0×10^6	
Average	134,500	125,100	15.8		18.1×10^6	

[a] Fabrication cycle consisted of heating at 400° F/hr from 1300–1700° F, holding 4 hr at 1700° F, cooling from 1700–1300° F in 90 minutes and cooling from 1300–1100° F in 60 minutes.
[b] Same as above [a] except specimens cooled half as rapidly.

The relationship between fracture mode and toughness of the joints is not understood clearly at present. Nevertheless, the substantial differences between failure modes of specimens bonded 1 or 4 hr at 1700° F is readily apparent, as is the difference between failure modes produced by rapid cooling versus controlled cooling. Additional effort is needed to firmly establish relationships between copper content, microstructure, fracture mode, and toughness in order to optimize processing parameters.

Tensile Properties of Ti-6A1-4V Base Metal and Butt Joints

Tensile properties of as-received (annealed) plate and sheet were determined (Table 4) to establish the effects of the bonding cycle on base metal properties and to provide a basis for comparison with tensile tests on butt-joints. The longitudinal and transverse properties of the annealed plate material were similar, indicating very little anisotropy. After exposure to the fabrication cycle, the plate material retained better than 90% of its original strength; tensile strength decreased approximately 4000 to 6000 psi, and yield strength decreased approximately 10,000 to 12,000 psi, with a corresponding increase in elongation and reduction in area for both orientations. These mild changes were probably caused by some grain growth during the exposure at 1700° F.

Longitudinal tensile properties of the annealed sheet were similar to those of the plate. Exposure of the sheet to the fabrication cycle decreased the longitudinal tensile strength approximately 8500 psi and the yield strength approximately 12,000 psi, with an increase in elongation and reduction in area. Thus, the effects of the fabrication cycle on tensile properties of the sheet and plate were similar. One set of longitudinal sheet specimens was exposed to a modified fabrication cycle in which the cooling rate was half the rate used for the standard fabrication cycle. This slower cooling rate simulated the slowest cooling rate to be expected for an assembly bonded in a ceramic tool. The tensile properties of the material exposed to this modified cycle were similar to the tensile properties obtained after the standard cycle.

Tensile properties of butt-joints were determined at temperatures from −150 to 900° F (Table 5). To obtain some measure of the thermal stability of the joints, room temperature tests were also conducted after a thermal exposure of 100 hours at 800° F in air. This thermal exposure approximated the same amount of Ti-6Al-4V/Cu diffusion as a thermal exposure of 30,000 hr at 650° F, a typical Mach 3 service life. Specimens identified with the prefixes B5 and B6 represent coupons taken from two separate bonding runs to provide some measure of the reproductibility of tensile properties. Specimens from both bonding runs exhibited similar tensile strengths.

Several specimens from both bonding runs failed in the joints when tested at room temperature and −150° F. However, the tensile-strength values were equivalent to strength values for specimens which failed in base metal. In addition, these values were very close to the tensile-strength values obtained for base metal exposed to the fabrication cycle (Table 4). Visual examination indicated that the joint fractures were relatively flat, whereas the base metal exhibited the more ductile "cup and cone" type of fracture.

A comparison of room temperature tensile values shows that the 100 hr

Table 5—Tensile Properties of Butt Joints[a]

Specimen number	Longitudinal tensile strength, ksi	Failure location
Tested at —150° F		
B5-9	169.6	Joint
-10	167.8	Base metal
B6-9	169.8	Base metal
-10	167.8	Base metal
Average	168.8	
Tested at room temperature		
B5-1	132.8	Base metal
-2	130.3	Base metal
-3	131.4	Base metal
B6-1	134.4	Joint
-2	131.4	Joint
-3	135.7	Base metal
Average	132.7	
Tested at room temperature after 100 hr at 800° F in air		
B5-11	137.6	Joint
-12	137.5	Joint
-13	137.0	Joint
B6-11	136.2	Base metal
-12	136.0	Base metal
-13	136.9	Base metal
Average	136.4	
Tested at 600° F		
B5-4	92.4	Base metal
-5	96.9	Base metal
-6	93.7	Base metal
B6-4	93.6	Base metal
-5	95.6	Base metal
-6	93.6	Base metal
Average	94.3	
Tested at 900° F		
B5-7	78.2	Base metal
-8	80.2	Base metal
B6-7	81.2	Base metal
-8	80.1	Base metal
Average	79.9	

[a] Fabrication cycle consisted of heating at 400° F/hr from 1300–1700° F, holding 4 hr at 1700° F, cooling from 1700–1300° F in 90 minutes and cooling from 1300–1100° F in 60 minutes.

thermal exposure appeared to increase strength slightly. All tensile tests at 600 and 900° F produced base metal failures. Thus, the tensile strength of butt-joints is at least as high as the tensile strength of base metal at temperatures in the range of —150 to 900° F.

Shear Strength of Lap Joints

The lap-shear specimens were joined with 30 mg Cu/in.^{2} rather than the 50 mg/in.^{2} used on butt-joints for two reasons. First, it was anticipated that the ability of sheet material to bend and conform would permit closer contact between faying surfaces than obtainable in rigid plate material. Thus, less eutectic liquid would be needed to fill the joint. Second, it was necessary to minimize formation of excess liquid which could form fillets

Table 6—Shear Strength of Lap Joints[a]

Specimen number	Lap-shear strength[b] (shear plus bending), ksi	Shear strength of annealed Ti-6Al-4V bar stock[c] (pure shear), ksi
Tested at —150° F		
L-4	89.7	
L-8	88.0	
L-14	87.0	
Average	88.2	105 (estimated)
Tested at room temperature		
L-3	74.6	
L-10	70.3	
L-16	70.8	
Average	71.9	90
Tested at 600° F		
L-2	58.0	
L-9	57.6	
L-13	59.3	
Average	58.3	71
Tested at 900° F		
L-5	56.1	
L-12	55.1	
L-18	51.0	
Average	54.1	58

[a] Lap joints fabricated from 0.062 in. sheet with 1-T overlap, using 30 mg. Cu/in.^{2} and a bonding cycle based upon heating at 500 F°/hr from 1300–1700° F, holding 1 hr at 1700 F° cooling from 1700–1300° F in 90 minutes, and cooling from 1300–1100° F in 60 minutes.
[b] All specimens failed by shear in joints.
[c] Data from Aerospace Structural Metals Handbook.

at the extremities of the joint and introduce errors into the determination of shear strength.

Metallographic examination of several lap joints showed no evidence of fillet formation. The bond lines contained approximately 10% voids, indicating insufficient liquid to effect a complete bond. Thus, fabrication of defect-free joints would require somewhat more copper. The maximum residual-copper content of these joints was estimated to range for 5 to 7% based upon examination of the microstructures.

Shear strength of the lap joints was determined at temperatures from —150 to 900° F (Table 6). The joints exhibited very high strength and good consistency. Data on the shear strength of annealed Ti-6Al-4V bar stock[4] are included for comparison with the lap shear data. Lap shear strength was 80 to 93% of the shear strength of Ti-6Al-4V bar stock from —150 to 900° F. The shear values for the bar stock represent pure shear; whereas the combined shear and bending loads in a lap shear test provide conservative shear strength values. Nevertheless, the shear strength of these imperfect lap joints was unusually high and well above shear strengths normally associated with

Table 7—Fatigue Properties of Base Metal Ti-6Al-4V and Butt Joints[a]

Specimen number	Maximum, stress ksi	Cycles to failure (× 10^6)	Failure location
As-received (annealed Ti-6Al-4V)			
T-2	122.4	0.426	Gage length
T-4	121.5	0.606	Gage length
T-6	112.8	0.506	Gage length
T-5	111.9	0.479	Gage length
T-7	104.0	1.447	Gage length
T-1	103.2	0.776	Threads
T-9	91.6	2.491	Gage length
T-8	91.5	3.824	Gage length
Ti-6Al-4V exposed to fabrication cycle[b]			
T-12	124.0	0.074	Gage length
T-10	121.4	0.279	Gage length
T-11	112.3	0.647	Gage length
Butt joints[b]			
B3-1	122.7	0.119	Base metal
B2-2	123.6	0.043	Joint
B2-4	122.0	0.056	Joint
B3-2	113.9	1.151	Base metal
B2-3	113.4	0.816	Base metal
B2-5	103.6	1.223	Base metal
B2-1	103.4	1.402	Base metal
B3-3	103.4	0.792	Joint
B2-6	99.5	2.676	Base metal
B2-7	93.9	2.654	Base metal
B3-6	89.2	0.494	Joint
B3-5	88.9	1.570	Base metal
B5-15	88.5	4.254	Base metal
B5-16	88.5	3.505	Base metal

[a] Sinusoidal, tension-tension fatigue tests at 1200 cpm with a stress ratio of 0.1.
[b] Fabrication cycle consisted of heating at 400° F/hr from 1300–1700° F, holding 4 hr at 1700° F, cooling from 1700–1300° F in 90 minutes, and cooling from 1300–1100° F in 60 minutes.

conventional brazing processes.

Fatigue Properties of Ti-6A1-4V and Butt Joints

Results of sinusoidal, tension-tension fatigue tests at 1200 cpm and a stress-ratio of 0.1 are recorded in Table 6 for as-received (annealed) Ti-6Al-4V, Ti-6Al-4V exposed to the fabrication cycle, and butt-joints. These results are also plotted in Fig. 13. Specimens identified with the prefixes B2, B3, and B5 represent coupons taken from separate bonding runs. No major differences were noted in the fatigue behavior of specimens from different bonding runs.

Of the 14 joints tested, 10 failed in the base metal and 4 failed in the joints. Specimen B3-6, tested at a maximum stress of 89.2 ksi, failed in the joint outside the scatter band, and the joint may have been defective. Specimen B3-3, tested at a maximum stress of 103.3 ksi failed in the joint after 792,000 cycles, which represented the low end of the fatigue scatter band. The remaining two joint failures, B2-2 and B2-4, occurred at maximum stress levels above 120 ksi.

Two of the as-received (annealed)

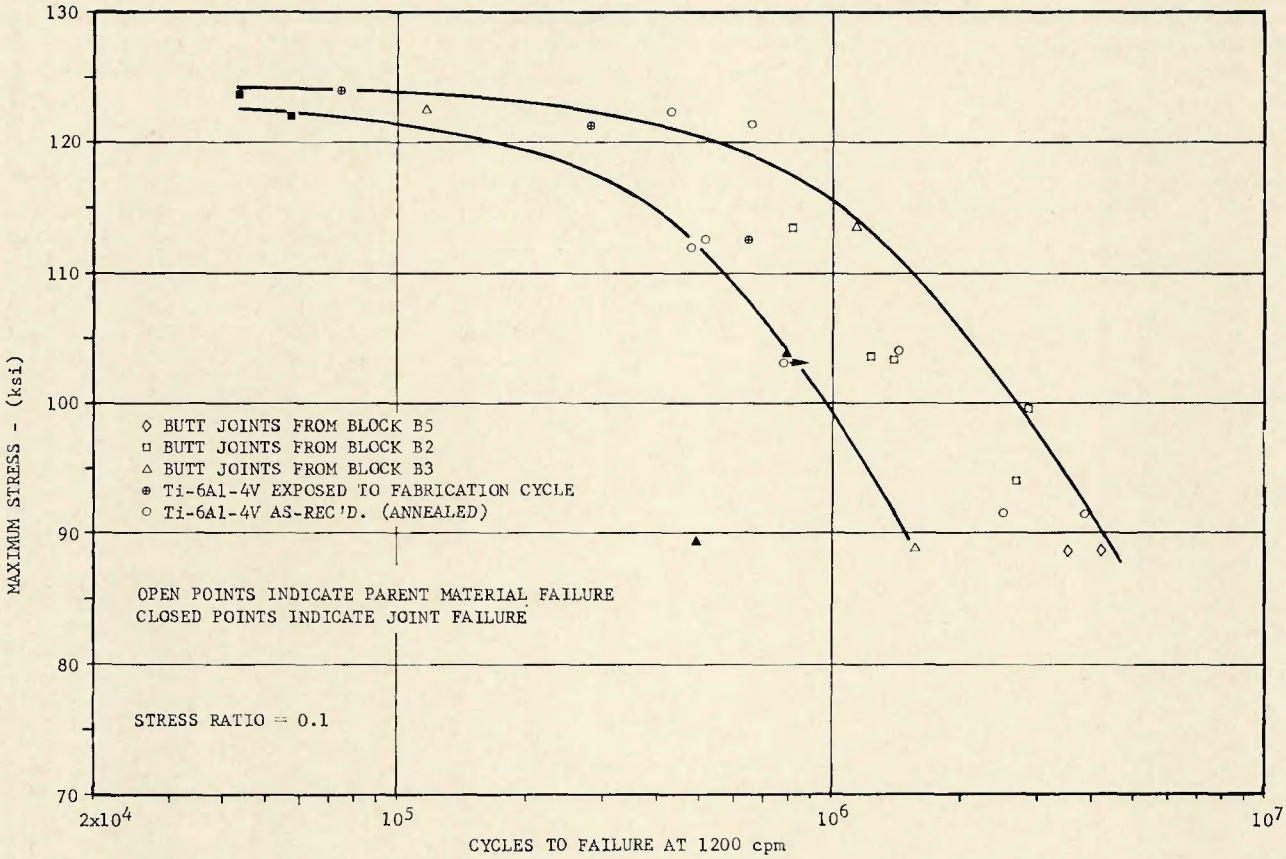


Fig. 13—Sinusoidal, tension-tension fatigue properties of base metal Ti-6Al-4V and butt joints

base metal specimens tested above a maximum stress of 120 ksi exhibited better fatigue life than the joints or base metal exposed to the fabrication cycle. This was probably related to the slightly higher strength of the annealed base metal. Below 120 ksi, the fatigue behavior of joints, base metal exposed to the fabrication cycle, and annealed base metal fell within the same scatter band, and no significant differences could be detected between the fatigue behavior of joints and base metal. At 10^6 cycles, the fatigue strength was approximately 100 to 115 ksi.

These tests indicate that the fatigue strength of smooth butt-joints is excellent and essentially equal to the fatigue strength of Ti-6A-4V base metal, particularly at stresses below 120 ksi. In addition, the results also show that the fabrication cycle does not reduce the fatigue strength of the Ti-6Al-4V base metal to any significant degree.

Stress Corrosion Behavior

Tests were conducted to determine threshold stress-intensity factors (K_{ISCC}) for stress corrosion in 3.5% NaCl at room temperature. All of the K_{ISCC} tests were conducted by periodically increasing the load on precracked WOL specimens in small increments until stress-corrosion crack growth occurred when the K_{ISCC} was exceeded. It was observed that stress-intensity factors as little as 10% above K_{ISCC} produced rapid crack growth and specimens failed within 1 to 2 hr. These failure times corresponded to crack-growth rates of approximately $^1/_8$ to $^3/_4$ in. per hour.

Table 8 shows that the K_{ISCC} values for the NOR-Ti-BOND joints are essentially equal to or slightly better than the K_{ISCC} values for as-received (annealed) base metal and for base metal exposed to the fabrication cycle. The ratio K_{ISCC}/K_{IC} provides a measure of the degree of stress-corrosion susceptibility produced by the salt water. The ratio for the joints is 0.77, which is substantially higher than the values of 0.42 and approximately 0.37 obtained for as-received base metal and base metal exposed to the fabrication cycle. Thus, the base metal is more susceptible to stress corrosion than the joints. In addition, the data show that the fabrication cycle had no major effect on the K_{ISCC} for base metal.

Several WOL specimens were examined by scanning electron factography to determine stress-corrosion failure modes for base metal and joints. As-received (annealed) and base metal exposed to the fabrication cycle exhibited similar stress-corrosion fracture modes. Figure 14 shows a typical stress-corrosion fracture for as-received base metal. The primary failure mode was transgranular quasi-cleavage with a small amount of secondary cracking. The stress-corrosion fracture mode of a joint produced by the standard fabrication cycle was also transgranular quasi-cleavage with transgranular cleavage through the alpha needles—Fig. 15.

In general, smooth specimens of titanium alloys are not susceptible to stress-corrosion in salt water. Several tests were conducted to determine stress-corrosion susceptibility of butt-joints in smooth specimens. In addition, one smooth base-metal specimen was tested after exposure to the fabrication cycle. After 1000 hr immersion in 3.5% NaCl at room temperature and under a tensile stress of 102,000 psi, the specimens showed no evidence of corrosion or stress corrosion. The specimens were then tensile tested at room temperature, yielding the results shown in Table 8. One specimen failed in the joint and the remaining specimens failed in the base metal. Tensile properties were basically the same as those obtained on specimens which were not exposed to a stress corrosion test (Table 4). Thus, smooth joints and base metal exposed to the fabrication cycle are essentially immune to stress corrosion in 3.5% NaCl at

Table 8—Stress Corrosion Behavior in 3.5% NaCl at Room Temperature of Ti-6Al-4V Base Metal and Joints

K_{ISCC} Determinations:

Specimen number	K_{ISCC}(ksi$\sqrt{\text{in.}}$)	K_{ISCC}/K_{IC}	Fabrication cycle and notes[a]
Joints (WR orientation of base metal)			
W-13	33.0 ± 0.4	0.73 ± 0.01	Joined using standard fabrication cycle.
W-14	36.6 ± 0.6	0.81 ± 0.01	
Average	34.8 ± 0.5	0.77 ± 0.01	
Base metal (WR orientation)			
TWR-3	34.0 ± 0.8	0.44 ± 0.01	As-received (annealed).
TWR-4	29.3 ± 0.8	0.39 ± 0.01	
Average	31.7 ± 0.8	0.42 ± 0.01	
TWR-9	34.3 ± 3.4	0.38 ± 0.04	Exposed to fabrication cycle.
TWR-10	32.0 ± 1.3	0.35 ± 0.01	
Average	33.2 ± 2.4	0.37 ± 0.03	

Smooth-specimen stress corrosion tests:
(Tensile properties of longitudinal specimens exposed 1000 hr at 102,000 psi in 3.5% NaCl)

Specimen number	Tensile strength, psi	Yield strength, psi	Elongation, %	Reduction in area, %	Young's modulus, psi	Failure location
Butt joints						
B3-7	136,000	125,000	15.0	27.7	17.1×10^6	Joint
B5-14	132,500	121,100	17.5	37.0	16.2×10^6	Base metal
B6-14	132,300	121,000	18.0	39.9	18.0×10^6	Base metal
Average	133,600	122,400	16.8	34.8	17.1×10^6	
Base metal exposed to fabrication cycle						
T-14	135,800	123,300	17.5	39.9	18.5×10^6	

[a] Fabrication cycle consisted of heating at 400° F/hr from 1300–1700° F, holding 4 hr at 1700° F, cooling from 1700–1300° F in 90 minutes, and cooling from 1300–1100° F in 60 minutes.

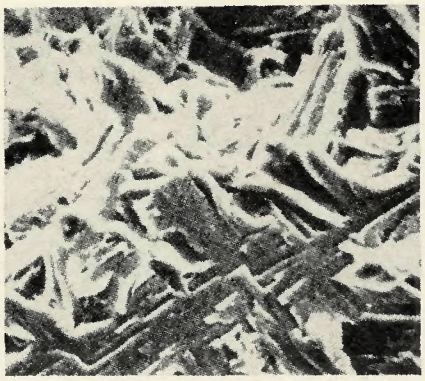


Fig. 14—Stress-corrosion fracture of annealed Ti-6Al-4V—specimen TWR-4. X1000 (reduced 41% on reproduction)

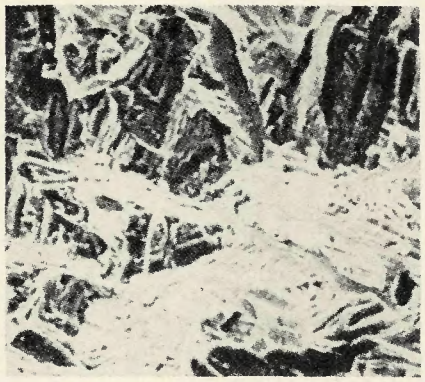


Fig. 15—Stress-corrosion fracture of joint—specimen W-13. X500 (reduced 41% on reproduction)

room temperature.

Fatigue and Corrosion Fatigue of WOL Specimens

The fatigue behavior of base metal and thin-film diffusion-brazed WOL specimens tested in air and in 3.5% NaCl is summarized in Fig. 16. The base metal exposed to the fabrication cycle exhibited the largest crack length at failure, followed by annealed base metal and then the joints. The crack lengths at failure were reasonably consistent with the fracture toughness values for the specimens. Because of the differences in fracture toughness, comparisons based upon total cycles to failure do not permit accurate comparisons between base metals and joints. A more reasonable basis for comparison is the total number of cycles required to grow a crack over a specific range.

Table 9 shows a comparison of fatigue and corrosion-fatigue behavior based upon the number of cycles required to grow a crack from 1.15 to 1.80 in. The base metal exposed to the fabrication cycle exhibited the best fatigue and corrosion-fatigue behavior, although the annealed base metal was almost equivalent. In 3.5% NaCl, the cracks grew approximately twice as fast as in air for the base metal in either condition.

For the joints fatigued in air, the crack started in the joint but propagated into and ran parallel to the joint but approximately $^1/_8$ in. from it. This problem was solved by side notches of $^1/_{16}$ in. radius and 0.025 in. depth along the joint. The fatigue and corrosion-fatigue behavior of base metal was somewhat superior to the behavior of the joints. It is noteworthy that fatigue tests in air on smooth butt-joints indicated comparable fatigue behavior for joints and base metal. However, in the presence of a sharp notch, the WOL specimen results indicate better fatigue behavior for base metal. For a joint in 3.5% NaCl, the crack grew approximately three times as rapidly as in air. Thus, the joint exhibited a higher relative susceptibility to corrosion fatigue than base metal. Additional studies are required to determine the effects of joint microstructure on fatigue and corrosion-fatigue behavior, which would provide information for optimizing processing parameters.

Several WOL specimens were examined by scanning electron fractography to determine fracture modes in fatigue and corrosion fatigue. As-received (annealed) and base metal exposed to the fabrication cycle exhibited similar fracture modes. Figure 17 shows that the annealed base metal fatigued in air exhibited a normal, transgranular, fatigue failure with a small amount of secondary cracking. The fracture mode under corrosion fatigue (Fig. 18) was transgranular quasi-cleavage with some secondary cracking. However, as the crack extended and the stress-intensity factor increased, the failure mode changed to transgranular cleavage with much less secondary cracking.

Under fatigue in air, the joint exhibited a normal transgranular fatigue fracture with a small amount of secondary cracking—Fig. 19. Under corrosion fatigue, the failure mode (Fig. 20) was transgranular quasi-cleavage with secondary cracking along interfaces between alpha needles and eutectoid as well as in the eutectoid. Some areas of transgranular cleavage through the alpha needles were also evident. Thus, the joints and base metal exhibited the same basic fracture modes.

It is interesting to note that the corrosion-fatigue failure modes for joints and base metals were basically the same as failure modes under stress corrosion. Thus, stress corrosion is apparently the dominant part of the failure mechanism operating in corrosion fatigue.

Analysis of Results

The thin-film diffusion-brazing porcess may be considered to be a combi-

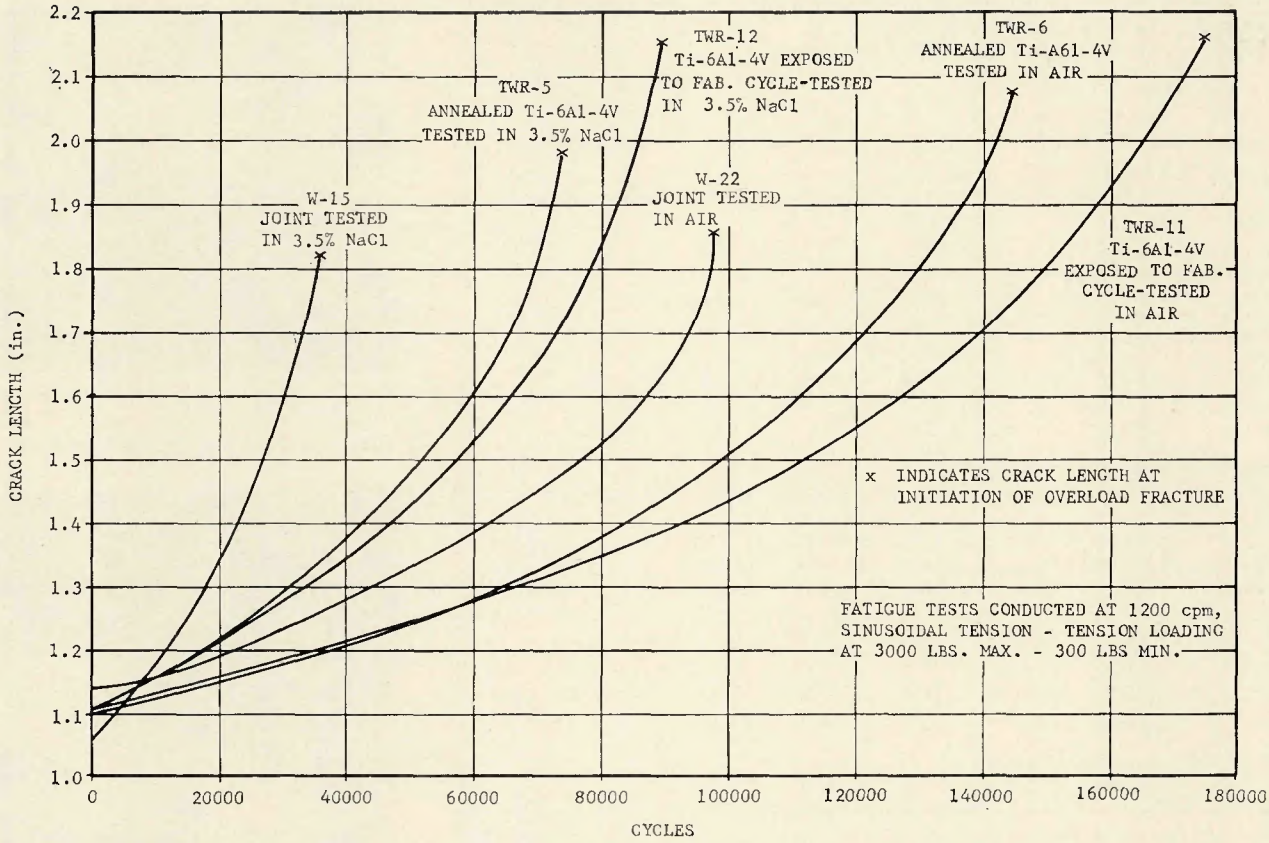


Fig. 16—Fatigue behavior of WOL specimens in air and 3.5% NaCl

nation of brazing and diffusion bonding. Thus, one would expect joint properties to be between the properties obtained by brazing and those obtained by solid-state diffusion bonding which should theoretically produce base metal properties The tensile tests on butt joints and shear tests on lap joints showed that the joints exhibited strength values which closely approximated the strength of base metal in the temperature range from −150 to 900° F. Thus, from a strength standpoint, thin-film diffusion-brazing closely approximated solid-state diffusion bonding.

Conventional brazing of titanium alloys has been performed using aluminum, Ti-47.5Zr-5Be and Ag-5Al filler metals. Titanium joints brazed with Ag-5Al provide joint strengths typical of what may be expected from conventional brazing. The lapshear strength of Ti-8Al-1Mo-1V brazed with Ag-5Al is typically 33,000 psi at room temperature and 18,000 psi at 600° F, [5] and similar values would be expected for Ti-6Al-4V joints brazed with this filler metal. In comparison thin-film diffusion brazing of Ti-6Al-4V provided lap-shear strengths of 72,000 psi at room temperature and 58,000 psi at 600° F. Thus, this joining process provides substantially better strength and strength retention at elevated temperature than conventional brazing.

Thin-film diffusion-brazed joints exhibited lower plane-strain fracture toughness (K_{IC}) than one would expect from joints produced by solid-state diffusion bonding. The K_{IC} of a joint containing approximately 7% maximum residual copper is approximately 45 ksi $\sqrt{\text{inch}}$ compared to a K_{IC} of approximately 78 ksi $\sqrt{\text{inch}}$ for annealed Ti-6Al-4V. For a given flaw size and specimen geometry, the load at which the flaw becomes critical is directly proportional to K_{IC}. Thus, the joint may be loaded to approximately 58% of the failure load for annealed base metal. While K_{IC} data are unavailable for conventional brazed joints, it is highly doubtful that toughness values would approach these values. In fact, the Ag-5Al joints containing no flaws can only be loaded to approximately 46% of the strength of Ti-6Al-4V based upon shear-strength comparisons at room temperature.

Table 9—Fatigue and Corrosion-Fatigue Behavior of Ti-6Al-4V and WOL Joints

Specimen number	Material	Cycles to grow crack[a]
Fatigued in air		
TWR-11	Ti-6Al-4V exposed to fabrication cycle[b]	130,000
TWR-6	Ti-6Al-4V annealed	110,000
W-22	Joint	88,000
Fatigued in 3.5% NaCl		
TWR-11	Ti-6Al-4V exposed to fabrication cycle[b]	69,000
TWR-6	Ti-6Al-4V annealed	60,000
W-15	Joint	28,000

[a] Number of cycles to grow crack from 1.15 to 1.80 in. in length under 1200 cpm sinusoidal tension-tension, maximum load 3000 lb, minimum load 300 lb.

[b] Fabrication cycle consisted of heating at 400° F/hr from 1300–1700° F, holding 4 hr at 1700° F, cooling from 1700–4100° F in 90 minutes and cooling from 1300–1100° F in 60 minutes.

In the presence of a fatigue crack, joints and base metal were susceptible to stress corrosion. The threshold stress-intensity factor for stress corrosion (K_{ISCC}) was 35 ksi $\sqrt{\text{inch}}$ for joints, 32 ksi $\sqrt{\text{inch}}$ for annealed Ti-6Al-4V, and 33 ksi $\sqrt{\text{inch}}$ for Ti-6Al-4V exposed to the fabrication cycle. Smooth butt joints and parent material specimens were immune to corrosion and stress corrosion in 3.5% NaCl at room temperature and at stress levels up to at least 102,000 psi. Thus, the stress corrosion behavior of the thin-film diffusion brazed joints was at least equivalent to the stress corrosion behavior of base metal.

The fatigue behavior of smooth butt joints was essentially equivalent to the fatigue behavior of annealed Ti-6Al-4V and Ti-6Al-4V exposed to the fabrication cycle, particularly at stress levels below 120,000 psi. Similarly, fatigue studies on pre-cracked

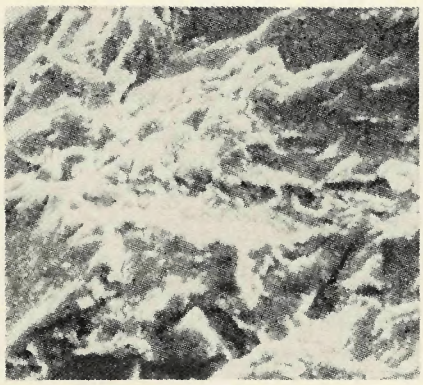
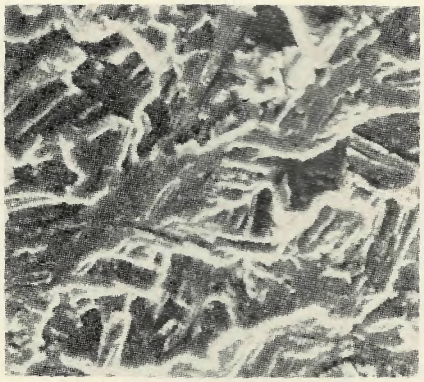


Fig. 17—Fracture surface of annealed Ti-6A1-4V fatigued in air—specimen TWR-6. X1000 (reduced 41% on reproduction)

plane-strain specimens tested in air showed that base metal exposed to the fabrication cycle exhibited approximately 18% better fatigue life than annealed Ti-6A1-4V. The fabrication cycle decreased strength but increased ductility and toughness which actually resulted in improved fatigue life. The joint exhibited approximately 80% of the fatigue life of annealed Ti-6A1-4V.

The fatigue lives of annealed Ti-6Al-AV and Ti-6Al-4V exposed to the fabrication cycle were reduced approximately 50% in 3.5% NaCl. The fatigue life of the joint was reduced approximately 68% in 3.5% NaCl. These limited data indicate that the joint exhibits a higher relative susceptibility to corrosion fatigue than base metal. These results are surprising in light of the fatigue and stress corrosion behavior of joints relative to base metal and the similarity of fracture modes for joints and base metal. Additional tests should be conducted to verify these results. Nevertheless, it is estimated from Fig. 16 that the joints and annealed Ti-6A1-4V would exhibit equal corrosion-fatigue lives if the joints were loaded at approximately 70% of the loading used for base metal.

Plane-strain fracture toughness is the controlling factor which limits the allowable loading on a thin-film diffusion brazed joint to approximately 58% of the load limit for annealed Ti-6A1-4V at room temperature. This limit is still a very high level compared to the limit one would expect for a conventional braze joint. It must also be appreciated that these limits are based upon properties of joints containing approximately 7% maximum residual copper. If the residual-copper content is reduced, the joint toughness would increase significantly with little change in strength, and higher loading limits could probably be realized. This can be accomplished readily by longer diffusion treatments in joints between



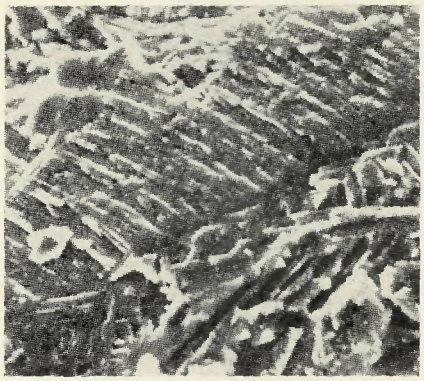


Fig. 18—Fracture surface of annealed Ti-6A1-4V fatigued in 3.5% NaCl—specimen TWR-5. A (top)—quasi-cleavage region; B (bottom)—cleavage region. X1000 (reduced 42% on reproduction)

heavy-gage material, but it would be difficult to achieve in thin-gage honeycomb structures where the amount of parent material available for diffusion is limited. Conversely, the thin-gage structures generally operate under plane stress conditions where toughness levels are higher than the toughness values under plane strain.

The results of this program indicate that most of the properties of thin-film diffusion-brazed joints equal or exceed base metal properties. One property, plane-strain fracture toughness (K_{IC}) is lower for joints but the threshold stress-intensity factor for stress corrosion (K_{ISCC}) is somewhat higher for

Fig. 20—Fracture surface of joint fatigued in 3.5% NaCl—specimen W-15. X1000 (reduced 38% on reproduction)

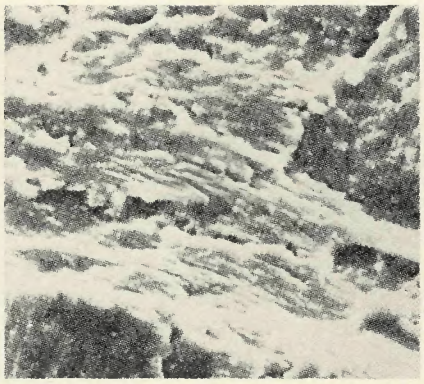


Fig. 19—Fracture surface of joint fatigued in air—specimen W-22. X2000 (reduced 38% on reproduction)

joints than for base metal. In addition, the thin-film diffusion-brazing process offers much better versatility and substantial manufacturing advantages over solid-state diffusion bonding. Thus, the process offers very attractive potential for producing high-integrity joints under reasonable manufacturing conditions, and it should find wide use for joining titanium-alloy structures.

Conclusions

Parameters were established for producing thin-film diffusion brazed joints across large faying surfaces between thick (rigid) Ti-6A1-4V plates. When faying surfaces were flycut flat and parallel within ±0.0005 in. with a surface finish of approximately 32 rms, 50 mg of copper sq in. of faying surface was required to produce a sound bond at a heating rate of 400° F/hr. Under these conditions, a minimum bonding pressure of approximately 50 psi was required when using copper foil and 100 psi was required when using electroplated copper.

Plane strain fracture toughness (K_{IC}) of the joints could be controlled in the range of 30 to 69 ksi $\sqrt{\text{inch}}$ by varying the bonding time from 1 to 12 hr at 1700° F to yield maximum residual-copper contents of 9.5% to 5.4%. These K_{IC} values ranged from approximately 40% to 90% of the K_{IC} for annealed Ti-6A1-4V. However, the threshold stress-intensity factor for stress corrosion (K_{ISCC}) in 3.5% NaCl at room temperature was approximately 35 ksi $\sqrt{\text{inch}}$ for joints, 32 ksi $\sqrt{\text{inch}}$ for annealed Ti-6A1-4V, and 33 ksi $\sqrt{\text{inch}}$ for Ti-6A1-4V evposed to the fabrication cycle. It should be recognized that these values of K_{IC} and K_{ISCC} for joints, pertain to cracks propagating strictly within the joint, not in the parent metal.

Smooth butt joints and Ti-6A1-4V

exposed to the fabrication cycle showed no evidence of corrosion or stress corrosion after a 1000 hr exposure at 102 ksi in 3.5% NaCl.

Base metal Ti-6Al-4V and joints were mildly susceptible to corrosion fatigue, with the joints showing somewhat higher susceptibility than the parent material.

In the presence of a pre-crack, the fatigue behavior of Ti-6Al-4V base metal was somewhat superior to the fatigue behavior of joints.

Fatigue strength of smooth butt joints at room temperature was essentially equivalent to the fatigue strength of Ti-6Al-4V base metal. Under sinusoidal, tension-tension fatigue at 1200 cpm and a stress ratio of 0.1, the fatigue strength of the joints was 100 ksi to 115 ksi at 10^6 cycles.

After exposure to the fabrication cycle, Ti-6Al-4V sheet and plate retained better than 90% of their original strength; tensile strength decreased 4000 psi to 8500 psi and yield strength decreased 10,000 psi to 12,000 psi, with a corresponding increase in ductility and toughness.

Tensile strength of the butt joints at −150 to 900° F was essentially equivalent to the tensile strength of Ti-6Al-4V exposed to the fabrication cycle.

In the temperature range from −150 to 900°F, shear strength of lap joints ranged from 88 ksi to 54 ksi. These unusually high values were 80% to 93% of the pure-shear strength of Ti-6Al-4V base metal.

Additional studies are desirable to optimize toughness, fatigue, stress corrosion, and corrosion-fatigue behavior of joints relative to microstructure and fabrication parameters. Nevertheless, present processing parameters produce joints which can be loaded to at least 58% of the load capacity of base metal Ti-6Al-4V. This loading limit is significantly higher than one would expect for conventional braze joints.

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"Unified Theory of Cumulative Damage in Metal Fatigue"

By Julien Dubuc, Bui Quoc Thang, Andre Bazergui and Andre Biron

A review is made of the different cumulative damage theories available in the literature. A new approach ("unified theory") is suggested which can be applied to stress-controlled or strain-controlled conditions, and which considers the order of application of different stress or strain levels. Comparison is made with a large number of test results using several different levels. The theory is also applied to some cases of random loading.

It is found that the proposed theory yields an improved agreement with experimental results, especially for cases where there is a large difference between levels.

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