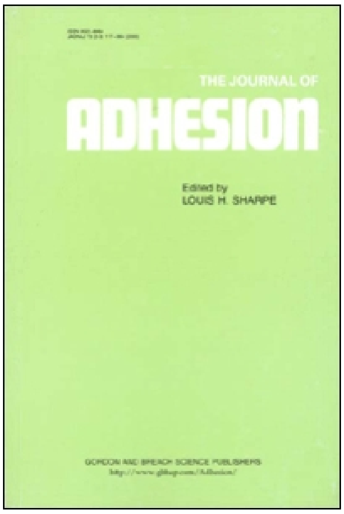

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Improved Fracture Toughness of a Titanium Alloy Laminate by Controlled Diffusion Bonding†

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The use of laminate composites containing a weak interface to increase the fracture toughness of high strength titanium alloys has been studied. Billets were fabricated from Ti-6Al-4V sheet material using a diffusion bonding process. Six billets were fabricated, each billet having an interface with different properties. Results indicate that toughness, as measured by the precracked Charpy test, may be increased when delamination or splitting of the bond occurs.

A simple model to predict the conditions necessary for delamination has been formulated. Correlations between the model and experimental results are made. The model can account for the effect of different base metal and interface material properties and thicknesses. It is seen that a thin, low yield strength interface material with a full strength diffusion bond to a high yield strength, fairly tough base metal leads to optimum composite toughness.

INTRODUCTION

There is a great need to develop high yield strength materials having high toughness. One traditional method of obtaining high toughness is through microstructural control or alloying. Another method is through the use of sub-macro laminated structures.[1] By using a high strength alloy with alternating layers of a weaker interfacial material, a practical "composite" material which can give superior fracture behavior may be obtained. This report presents the results of work done in evaluating the mechanical properties of titanium 6Al-4V multi-laminate materials.

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The concept of using a laminated structure to increase toughness is based upon the relaxation of triaxial constraint which is developed ahead of the crack tip. If the interface is sufficiently weak, the triaxial stresses or strains will cause delamination or separations ahead of the crack.

There are basically three possible orientations of the interface in relation to a propagating crack (Figure 1). In the arrester orientation the delamination would effectively blunt the crack. Gross yielding of the material occurs before ductile failure; linear elastic fracture toughness criteria are then not applicable.

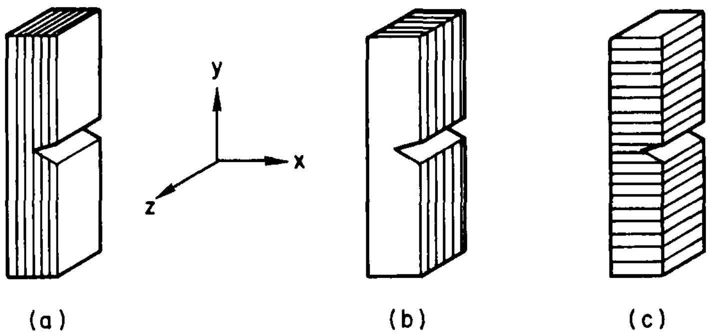


FIGURE 1 Possible orientation of the interface in relation to the propagating crack: (a) crack arrester, (b) crack divider, (c) crack enhancer.

In the divider orientation, delamination would cause the material to fail in the "plane stress" mode and toughness would again be increased. This results from the fact that the fracture toughness of thin sheets, K_c, is much greater than the plane strain fracture toughness, K_{Ic}, unless the sheets are very thin. However, if the crack lies in the weak interface plane, (enhancer orientation, Figure 1), the fracture toughness may be lower than that of base material. Even in uncracked material, the transverse strength of a macro-laminated composite will be somewhat less than that of the base material alone. The key question in composite material development is whether increased longitudinal toughness through crack arrest and crack division can be achieved without severe loss in transverse strength, for a given application. Optimization of the laminate and interface characteristics will allow for increases in toughness in the arrester and divider cases without significant loss of load carrying capabilities in the direction perpendicular to the plane of the interface (enhancer geometry).

The present work was designed to determine quantitatively the degree of "weakening" that is required in the interface in order to achieve high toughness in the two other orientations. The studies were conducted to determine whether there is a potential for using this approach to improve toughness in high strength titanium alloys.

MATERIALS AND FABRICATION

The material selected for the study was mill annealed titanium 6Al-4V sheet (AMS 4911) 0.040 inch in thickness. This thickness was chosen so

TABLE I

Basic properties and diffusion bonding parameters of the billets

Billet No.	Interleaf material	As fabricated tensile strength of bond Avg. of two specimens	Time (hrs.)	Temperature (°F)	Pressure (psi)
1	NONE (Full strength)	No tests made	5	1700	2000
2	NONE (Microvoid)	68 ksi	0.75	1350	5000
3	0.005″ CP Ti foil	132 ksi	5	1700	2000
4[a]	NONE (Microvoid)	75 ksi	1.5	1400	5000
5	0.010″ CP Ti foil	125 ksi	5	1700	2000
6	0.0015″ Al foil	38 ksi	3	1000	6000

[a] Billet No. 4 specimens were given a thermal treatment of ½ hour at 1600°F with no applied pressure which raised the bond strength to 142 ksi.

that a representative number of sheets (ten) would be present in the small (0.4 inch square), inexpensive Charpy specimens used in the work.

All billets were fabricated using a diffusion bonding process. The titanium sheet, retort, and tooling were prepared using established cleaning procedures. All diffusion bonding was conducted under vacuum of less than 10^{-4} torr. Time, temperature and pressure depended upon bond strength desired or interleaf material used. The parameters used for each of the billets are shown in Table I.

Six billets were fabricated. The as-fabricated tensile strength of the interface for the different billets is also shown in Table I. Billet size was nominally 7 inches wide by 7 inches long by 2.4 inches thick. Billet No. 1 was fabricated using the best established techniques to provide base line fracture behavior for comparison with the controlled bond strength billets. An ultimate tensile strength (UTS) of 135 ksi perpendicular to the bond line

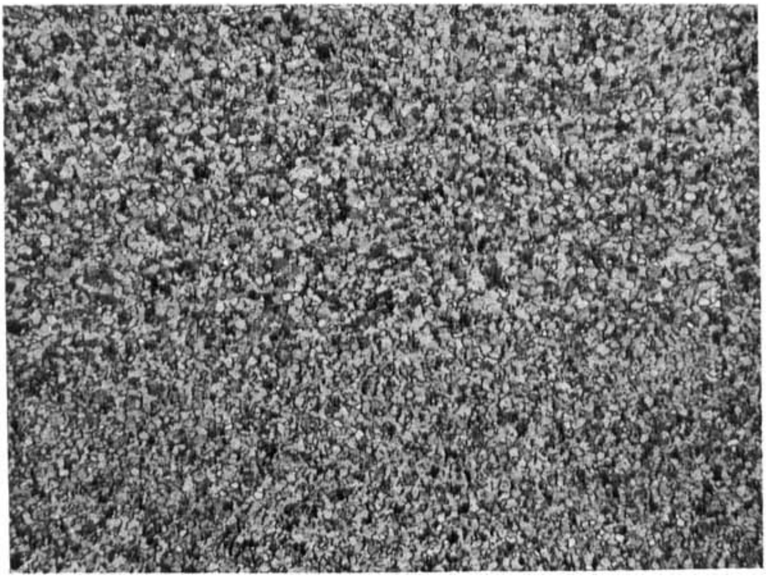


FIGURE 2 Typical microstructure of Billet No. 1 (100X).

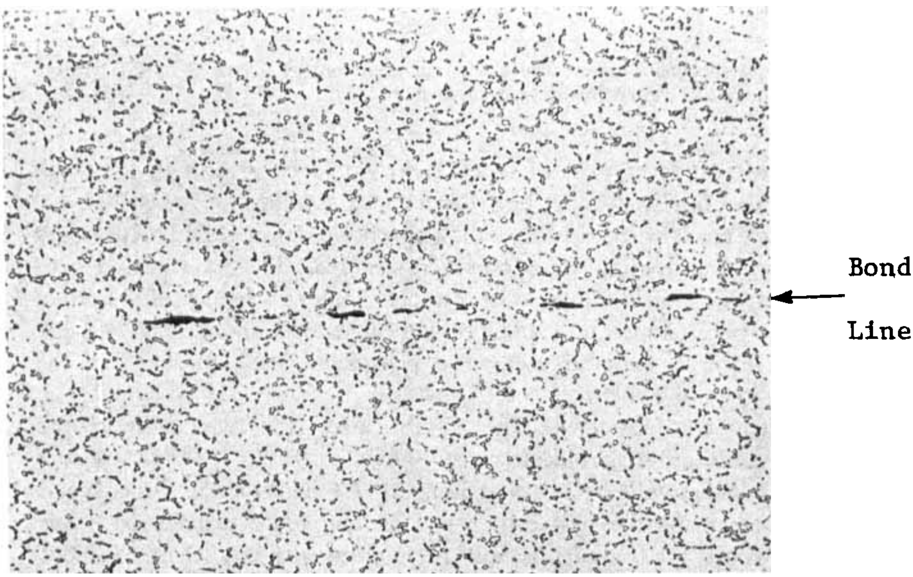


FIGURE 3 Typical microstructure of Billet No. 2 (500X).

was obtained. Figure 2 shows the typical microstructure of Billet No. 1. Billet No. 2 was fabricated with a controlled porosity bond of 68 ksi UTS perpendicular to the bond line. In Figure 3 the microstructure of this billet is shown and the remaining microvoids at the bond are easily visible. Billet No. 3 has a 0.005 inch 75A grade commercially pure titanium foil (AMS 4901) interleaf between the lamella. The diffusion bonding process was such that a full strength bond was obtained and the UTS of this interleaf was 132 ksi. The microstructure of billet No. 3 is shown in Figure 4.

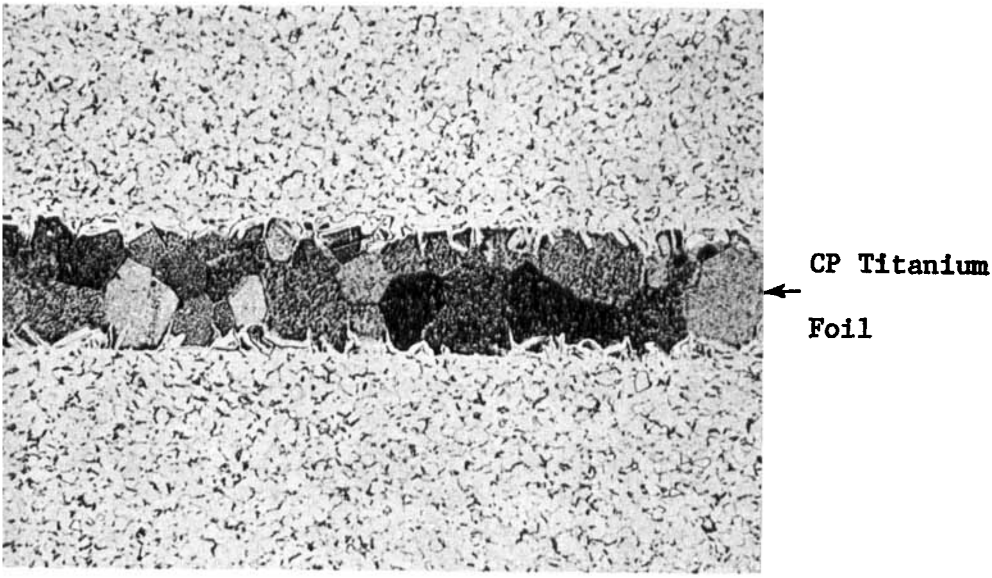


FIGURE 4 Typical microstructure of Billet No. 3 (100X).

Billet No. 4 was to have a controlled porosity bond with a UTS of 110 ksi. After fabrication, tensile tests indicated that the strength of the bond was only 75 ksi. A treatment of 30 minutes at 1600°F was used on the material in an attempt to obtain the desired bond strength. Tensile tests on material treated in this manner showed the resulting bond strength to be 142 ksi.

Billet No. 5 incorporated a commercially pure titanium foil (AMS 4901). The foil thickness was 0.010 inch and the resulting interface UTS was 125 ksi. Billet No. 6 was fabricated using a commercially pure aluminum foil 0.0015 inch in thickness as an interleaf material (1100 alloy). Tests showed the resulting interleaf to have a UTS of 38 ksi.

TESTING PROCEDURES

Because of the limited nature of the program, it was desirable to use a relatively small, inexpensive specimen for the determination of fracture toughness. Therefore, precracked ($\rho \approx 0.0001$) Charpy specimens (Figure 5) were used in this study. This specimen does not meet ASTM specimen criteria and the results are not considered "valid" K_{Ic} values. However, the values are valuable for comparing the relative toughness characteristics of the materials evaluated. From each billet eighteen specimen blanks were machined—six from each of the three orientations shown in Figure 1. The rough specimens were then heat treated prior to final machining.

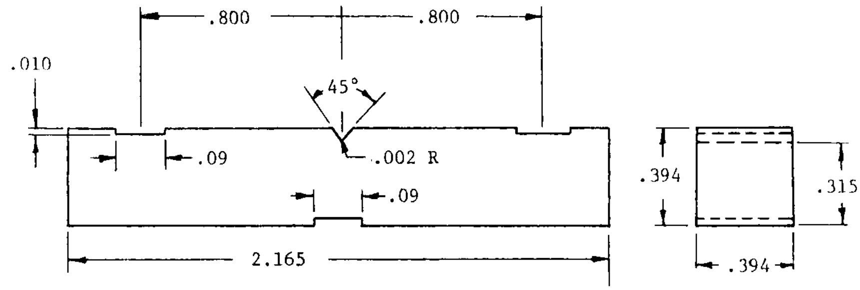


FIGURE 5 Charpy specimen used for fracture toughness testing.

Three material conditions were used in this study. Except for specimens from Billet No. 6, half of the specimens were given an annealing heat treatment while the remaining half were solution treated and aged. The annealing treatment consisted of $1\frac{1}{2}$ hours at 1400°F followed by furnace cooling. The vacuum present during this treatment was better than 10^{-4} torr. Solution treatment was performed at 1600°F for twenty minutes in air. They were then water quenched. The specimens were then aged six hours at 950°F while a roughing pump maintained a vacuum of approximately 0.1 torr, to halt the possibility of further oxidation. The specimens were then removed from the furnace and allowed to air cool.

Since both the annealing and solution heat treatments involve temperatures above the melting point of aluminum, it was not possible to heat treat Billet No. 6 material to these conditions. Specimens from this billet were machined from the as-fabricated material. Since the alloy used for fabrication was in the mill annealed condition, these specimens would be in an annealed condition, although the final anneal was not made.

Final machining of the specimens was done after heat treatment to insure the removal of the surface oxidation which occurred during the heat treatment. After final machining the specimens were precracked in three point bending. Approximately 50,000 cycles were used to obtain a fatigue crack of about 0.040 to 0.060 inch in depth.

The fracture toughness testing of the precracked Charpy specimens was conducted in slow, three point bending on an Instron tensile testing machine. The tests were run at room temperature and with a cross head speed of 0.1 in./min. Three specimens for each of the three possible crack orientations for both heat treat conditions were tested for a total of eighteen specimens per billet, except for Billet No. 6 where only nine specimens were tested.

RESULTS

A summary of the fracture toughness results is shown in Table II. Where delamination occurred in the arrester geometry, gross yielding and crack blunting followed, and the plane strain fracture toughness could not be determined.

TABLE II

Fracture toughness results from precracked Charpy specimens

		Fracture toughness ksi $\sqrt{in.}$	
Billet number	Orientation	Annealed condition	Heat treated condition
	Arrester	76.4	51.1
1	Divider	82.5	55.8
	Enhancer	75.7	37.9
	Arrester	arrest	arrest
2	Divider	73.4	60.7
	Enhancer	52.7	44.3
	Arrester	82.5	arrest
3	Divider	84.2	54.5
	Enhancer	59.0	32.7
	Arrester	74.2[a]	arrest
4	Divider	82.6	63.7
	Enhancer	73.5	33.9
	Arrester	82.2	74.3
5	Divider	83.6	59.4
	Enhancer	60.0	45
	Arrester	arrest	—
6	Divider	68.8	—
	Enhancer	48	—

Billet No. 1 was used as a control for comparison with the other billets containing the various interfaces. The variation in toughness with orientation in Billet No. 1 is due to the anisotropic nature of the sheet material used in the fabrication.[2]

Delamination in the arrester geometry occurred in both annealed and heat treated material of Billet No. 2. Shearing of the interface also occurred and was especially prevalent in the heat treated material.

In the divider geometry delamination and splitting was found although the toughness was not increased for the annealed material with this orientation. The fracture surface of the enhancer orientation was very flat and fracture occurred in the interface.

Delamination in arrester orientation of Billet No. 3 only occurred in the heat treated material. The divider orientation did not split in the annealed condition and the fracture surface was not completely shear lip in the heat treated specimens. Again the surface of the enhancer geometry was very flat.

One annealed arrester specimen delaminated for Billet No. 4 while all three heat treated specimens with this orientation delaminated. The divider behavior was similar to Billet No. 3. Little shear lip was observed in the annealed material while the fracture surface of the heat treated material was virtually 100% shear. Again, in the enhancer orientation there was a decrease in the fracture toughness with decrease in bond strength.

No delamination occurred in Billet No. 5 and the fracture surfaces appeared similar to those of Billet No. 1. Billet No. 6 was tested only in the as-fabricated (annealed) condition. Delamination and splitting occurred in both arrester and divider orientation in this billet. The enhancer fracture surface was flat and fracture occurred through the foil interleaf. Figures 6 and 7 show typical fracture appearances when delamination and splitting take place.

Figures 8 and 9 show values of fracture toughness as a function of the relative bond strength for the two conditions tested. The relative bond strength was found by taking the ratio of the bond strengths for the billets to the bond strength of Billet No. 1. It was assumed that the bond strength for Billet No. 1 was 145 ksi, which is a typical handbook value.

Figure 8 shows that in the annealed condition with arrester geometry there is a general increase in fracture toughness with decrease in bond strength. The divider orientation shows a decrease in toughness in the weaker bonds. The general trend in the enhancer geometry is towards a decrease in fracture toughness with decrease in bond strength.

In the heat treated material, delamination occurred in the arrester geometry for all specimens except those with the thicker titanium foil interleaf (Billet No. 5). There was little change in toughness with bond strength for

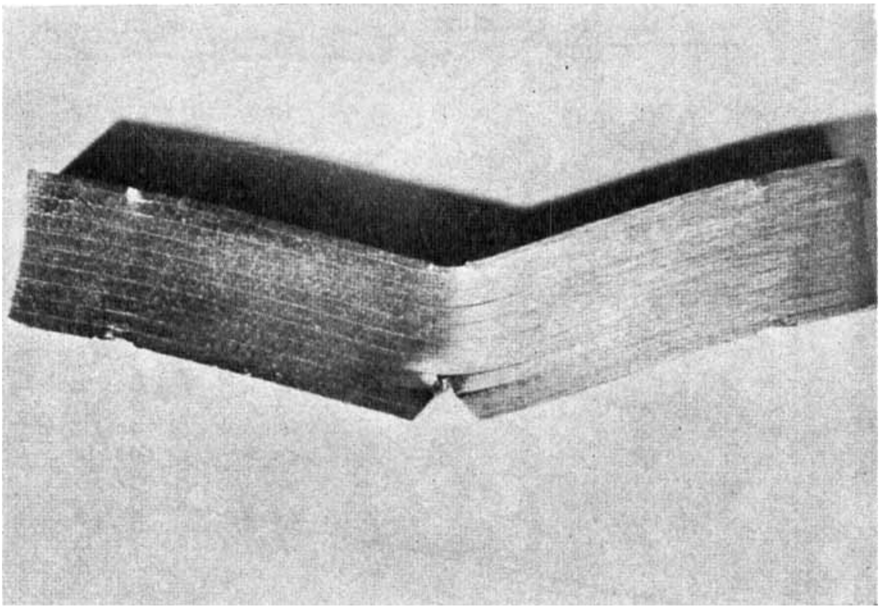


FIGURE 6 Delamination and crack arrest in arrester orientation of Billet No. 6.

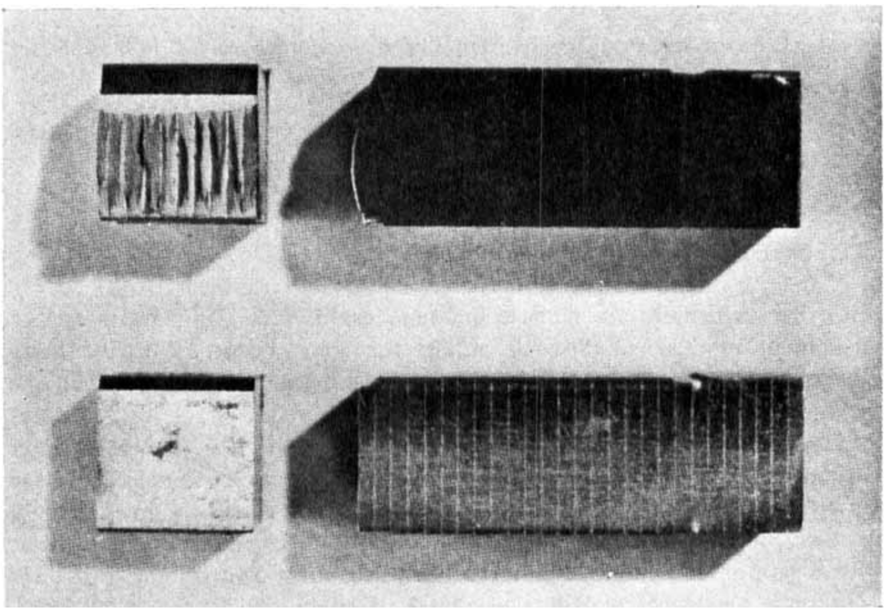


FIGURE 7 Fracture appearance of divider (top) and enhancer orientation of Billet No. 6.

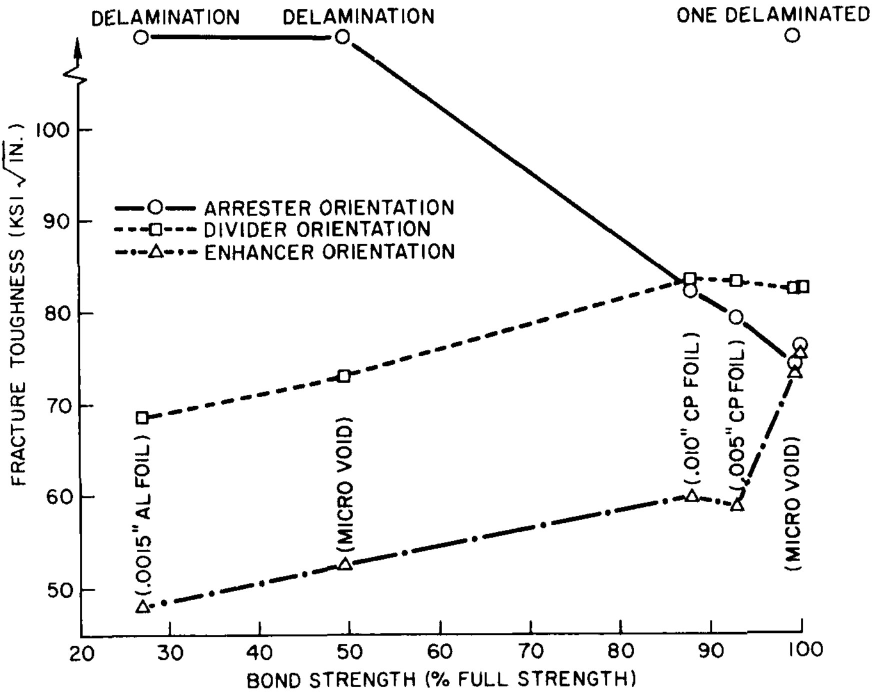


FIGURE 8 Fracture toughness vs. relative bond strength for material in the annealed condition.

the divider orientation. Again, in the enhancer orientation there was a decrease in the fracture toughness with decrease in bond strength.

THEORETICAL CONSIDERATIONS

In order to achieve an increase in toughness in these laminate materials, delamination ahead of the propagating crack must occur. Figure 10 shows the steps which lead to delamination and splitting of the interface in the arrester orientation. Triaxial constraint ahead of the crack tip produces transverse stress σ_{xx} and σ_{zz} in this region (Figure 10a). In the arrester orientation the transverse stress σ_{xx} acts as a tensile stress across the interface. As discussed by Saxton *et al.*[3] this tensile stress produces a shear stress along the bond line which in turn produces a second hydrostatic stress field in the interface material. This secondary hydrostatic stress field reinforces the effect of the transverse stresses caused by the crack itself. This combined triaxial stress state produces localized yielding around

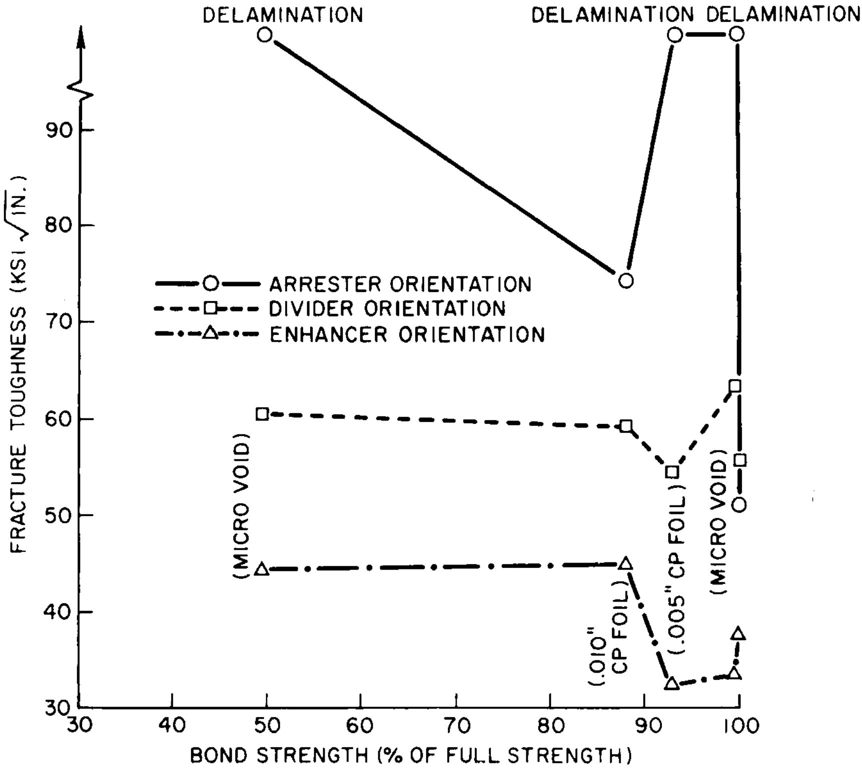
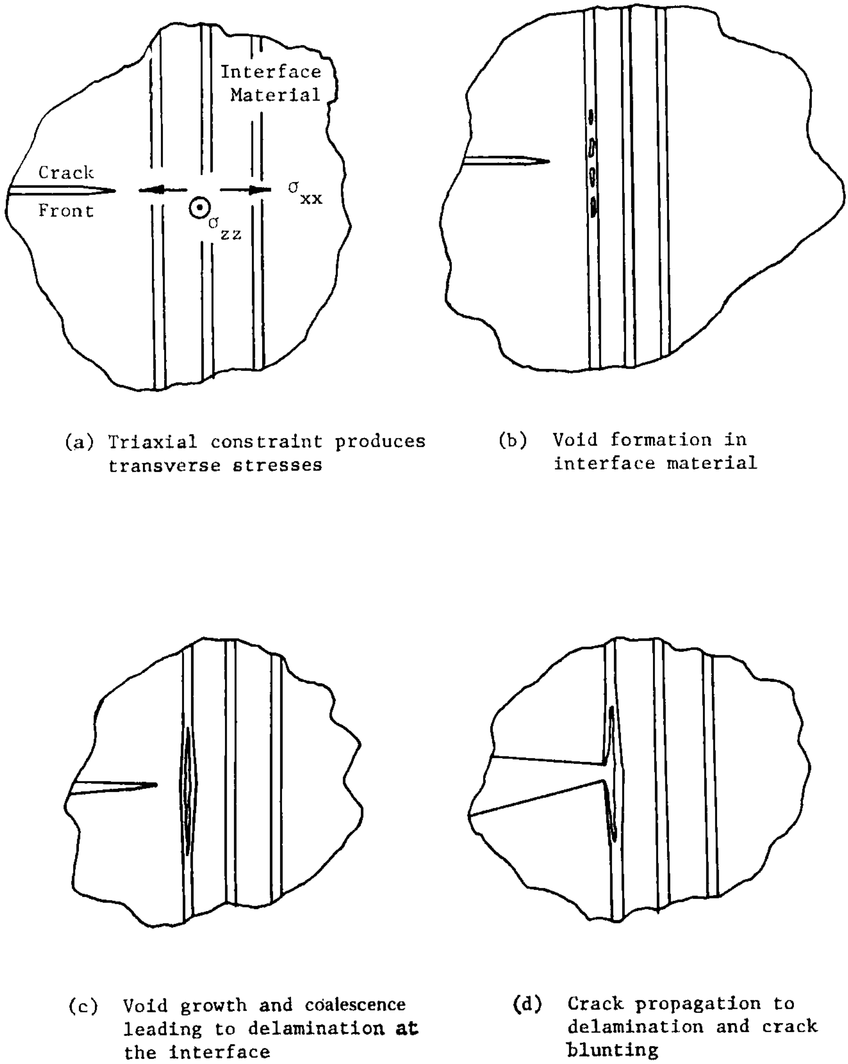


FIGURE 9 Fracture toughness vs. relative bond strength foı material in the heat treated condition.

void-nuclei present in the interface material (Figure 10b). Fracture occurs at the interface ahead of the advancing crack by the growth and coalescence of the voids (Figure 10c). The primary crack propagates to the delamination where it is effectively blunted (Figure 10d). Further crack propagation requires extensive gross yielding.

The situation is similar in the divider orientation (Figure 11). In this case the principal transverse stress which leads to splitting is the σ_{zz} stress ahead of the crack tip. The splitting of the interface causes the laminates to behave as the sum of many thin sheets. In this case, the fracture toughness approaches the plane stress toughness K_c. Since K_c can be a factor of 2 or 3 greater than K_{Ic} in high yield strength materials, delamination in the divider orientation can lead to a substantial increase in toughness, even though we have not observed these increases in our investigation.

The splitting of the interleaf type interface takes place in the interleaf by void formation and coalescence. In the microvoid interface, splitting takes



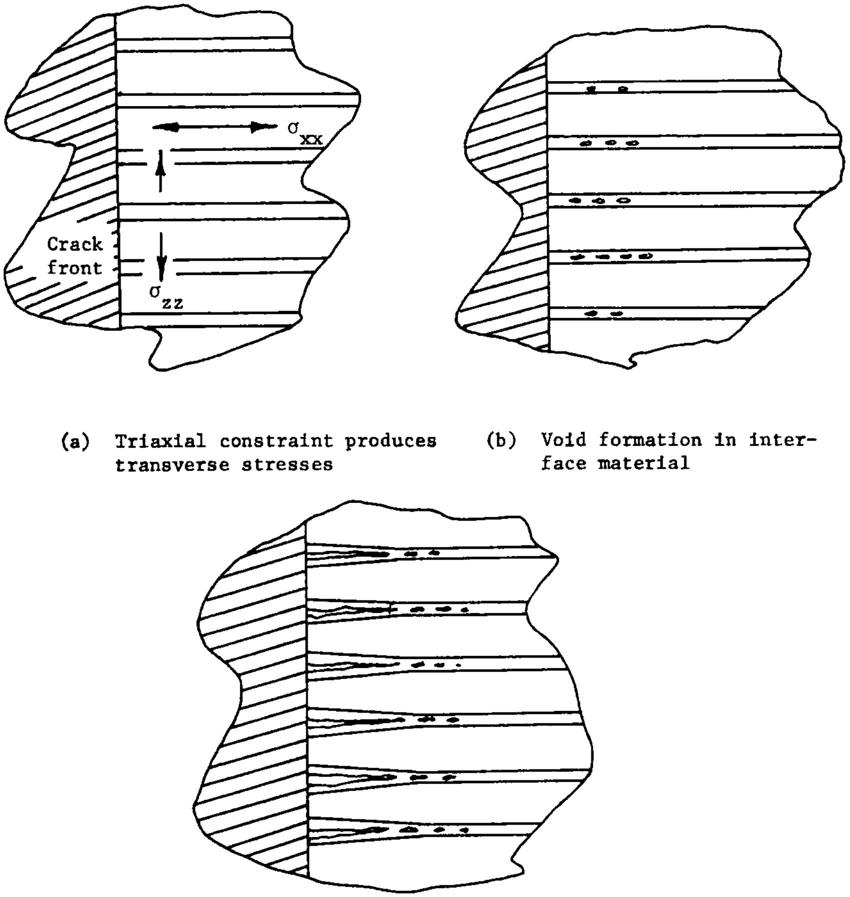
(a) Triaxial constraint produces transverse stresses

(b) Void formation in interface material

(c) Void growth and coalescence leading to delamination at the interface

(d) Crack propagation to delamination and crack blunting

FIGURE 10 Steps that lead to delamination in the arrester orientation.



(a) Triaxial constraint produces transverse stresses

(b) Void formation in interface material

(c) Void growth and coalescence leading to delamination and splitting of material into thin sheets

FIGURE 11 Steps that lead to delamination in the divider orientation.

place at the bond line by coalescene of the microvoids which are already present. This is seen in Figure 12 which shows a typical delamination surface for the divider orientation. The surface appearance is dimpled, indicating a ductile type fracture.

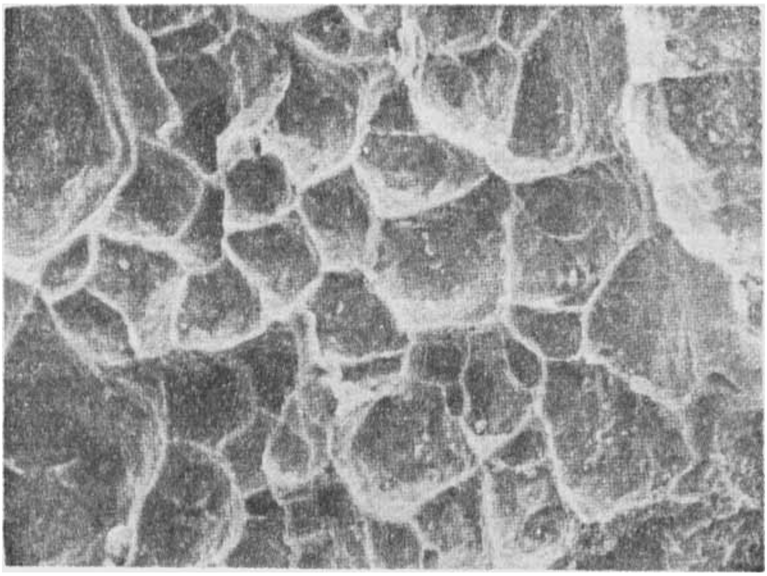


FIGURE 12 Fracture appearance of interface in the divider orientation of Billet No. 6 (430X).

From linear elastic fracture mechanics, it is known that crack propagation under plane strain tensile loading will occur when the stress intensity factor of the crack, K_I, equals the fracture toughness of the material K_{Ic}. The stress intensity is of the form

$$K_I = \sigma \sqrt{\pi a \alpha} \tag{1}$$

where σ is the applied stress, a is the crack length, and α is a parameter which depends upon the geometry of crack and specimen.[4] Thus, for fast fracture at a stress $\sigma = \sigma_F$

$$K_{Ic} = \sigma_F \sqrt{\pi a \alpha} \tag{2}$$

Therefore, to achieve an increase in toughness by use of laminated composites delamination and splitting must occur before the stress intensity factor equals the fracture toughness K_{Ic} of the material.

Localized ductile fracture of a material occurs when the plastic strain reaches some critical strain, ε_f. This criterion would apply for laminate composites where the interface failure has been shown to occur by dimple rupture. Thus for delamination to occur, the local strain in the interface must reach the critical value, ε_f, before $K_I = K_{Ic}$.

McClintock has shown that this critical fracture strain is inversely proportional to the degree of triaxiality in the material.[5] West *et al.*[6] studied the fracture of thin braze joints under uniaxial loading (isostress conditions). It was found that during loading, constraint of the interface produces a triaxial stress state in the joint material. This hydrostatic stress state was found to be a function of joint thickness, increasing as the thickness was decreased.

Laminate materials behave similarly. Consider the divider orientation, where the transverse stress σ_{zz} is set up across the lamella and a triaxial stress state is formed at the base metal/lamella interface. This high degree of combined triaxiality will decrease the critical strain necessary for fracture in accordance with the McClintock model, which predicts a strong inverse dependence of fracture strain on degree of triaxiality.

The transverse stress perpendicular to the plane of the interface (σ_{xx} in the arrester orientation, σ_{zz} in the divider orientation) will produce strains in both the base alloy (ε_b) and the interface material (ε_i). The stress will be the same in both materials but the plastic strains ε_i in the softer interface material are much greater than in the base material (Figure 13). Fracture of the interface will occur if the strain ε_i reaches the critical strain for fracture

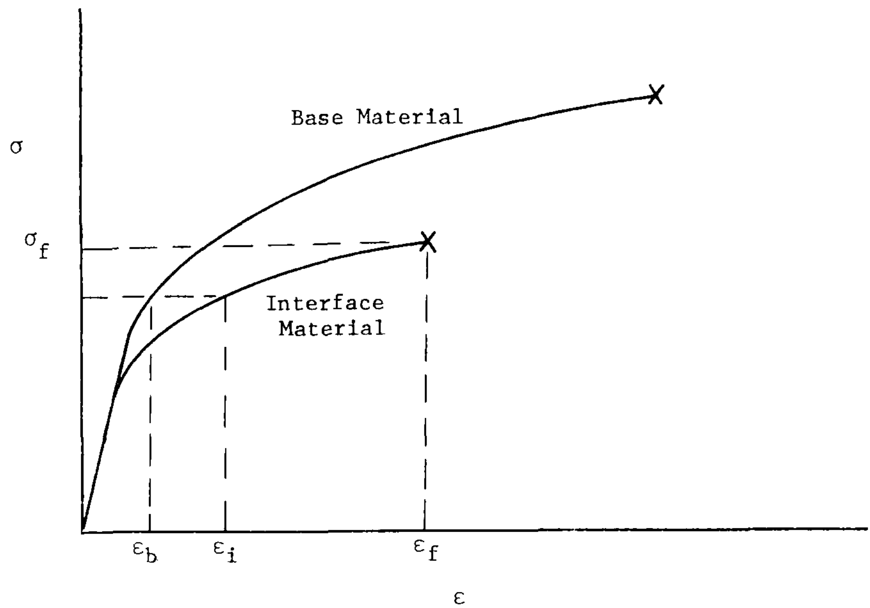


FIGURE 13 Schematic stress-strain curve for base and interface materials.

ε_f before $K_I = K_{Ic}$. This requires that the transverse stresses (σ_{xx} or σ_{zz}) reach the fracture stress of the interface σ_f before K_I reaches K_{Ic} for fast crack propagation.

Prior to general yield, the maximum transverse stresses ahead of a notch in a Charpy specimen occur at the elastic/plastic interface and are given by:[4]

$$\sigma_{xx} = \sigma_Y[\ln(1 + R/\rho)] \tag{3}$$

$$\sigma_{zz} = \nu\sigma_Y[1 + 2\ln(1 + R/\rho)] \tag{4}$$

where σ_Y is the yield strength of the material, ν is Poisson ratio, R is the plastic zone size and ρ is the notch root radius. For sharp cracks we can set $\rho = \rho_o$. From linear elastic fracture mechanics:[7]

$$R/\rho = \frac{R}{\rho_o} = \frac{1}{3\pi\rho_o}\left(\frac{K_I}{\sigma_Y}\right)^2 \tag{5}$$

and substituting in Eqs. (3) and (4):

$$\sigma_{xx} = \sigma_Y\left[\ln\left(1 + \frac{1}{3\pi\rho_o}\left(\frac{K_I}{\sigma_Y}\right)^2\right)\right] \tag{6}$$

$$\sigma_{zz} = \nu\sigma_Y\left[1 + 2\ln\left(1 + \frac{1}{3\pi\rho_o}\left(\frac{K_I}{\sigma_Y}\right)^2\right)\right] \tag{7}$$

The criterion for delamination can then be written in terms of the stress intensity factor

$$\sigma_{xx} = \sigma_f = \sigma_Y\left[\ln\left(1 + \frac{1}{3\pi\rho_o}\left(\frac{(K_I)_s}{\sigma_Y}\right)^2\right)\right] \tag{8}$$

$$\sigma_{zz} = \sigma_f = \nu\sigma_Y\left[1 + 2\ln\left(1 + \frac{1}{3\pi\rho_o}\left(\frac{(K_I)_s}{\sigma_Y}\right)^2\right)\right] \tag{9}$$

where $(K_I)_s$ is the critical stress intensity factor required for splitting (delamination), i.e., that necessary to obtain the critical fracture strain ε_f in the interface material. Delamination will occur before fast crack propagation, and increased toughness will result when

$$(K_I)_s < K_{Ic}$$

From Eqs, (8) and (9) it is possible to solve for the critical stress intensity factor required for delamination rather than brittle fracture:

$$(K_I)_s = [3\pi\rho_o\,\sigma_Y^2\,[\exp[\sigma_f/\sigma_Y] - 1]]^{1/2}\ \text{(arrester)} \tag{11}$$

$$(K_I)_s = [3\pi\rho_o\,\sigma_Y^2\,[\exp[(\sigma_f - \nu\sigma_Y)/2\nu\sigma_Y] - 1]^{1/2}\ \text{(divider)} \tag{12}$$

PARAMETERS WHICH AFFECT DELAMINATION

From the previous discussion, there are six basic physical and mechanical properties which can affect the delamination process.

1 Fracture toughness of the base alloy, $(K_{Ic})_b$

An increase in K_{Ic} for the base alloy will allow for a higher K_I before fast fracture will initiate. The result is an increase in the transverse stress and a corresponding increase in the interfacial strain. Therefore, a high K_{Ic} will increase the possibility of delamination, if all other parameters in Eqs. (11) and (12) are held constant.

2 Yield strength of the base alloy, $(\sigma_Y)_b$

The yield strength of the composite may be approximated by the base metal yield strength if a thin interleaf is used, as in the present experiments. Equations (11) and (12) indicate that an increase in σ_Y tends to both increase (pre-exponential term) and decrease (exponential term) $(K_I)_s$. Since the exponential term dominates, $(K_I)_s$ decreases with increasing yield strength σ_Y.

In high strength materials, the fracture toughness K_{Ic} also decreases with increasing σ_Y. Since only one base alloy was used in the present work, it was not possible to study independently the effect of yield strength or fracture toughness of the base alloy. Results obtained from Billet No. 3 with the 0.005 inch titanium foil show the combined effect of these two base alloy material properties. Whereas no delamination occurred in the annealed material $[(K_{Ic}/\sigma_Y) = 0.55]$, the heat treated specimens with a higher yield strength and slightly lower K_{Ic} did delaminate $[(K_{Ic}/\sigma_Y) = 0.35]$. These results would indicate that increases in σ_Y favor delamination if K_{Ic} is not lowered too greatly.

3 Sheet thickness of the base alloy, (t_b)

The thickness of the base alloy has little effect on whether the delamination process will occur in the laminated material. However, the fracture toughness K_c is a function of the thickness of the material. Fracture toughness increases to a maximum at a thickness t^*, then decreases with further increase in thickness until the plane strain value K_{Ic} is reached (Figure 14). The initial increase in thin sheets has been attributed by Hahn and Rosenfield[8] to an increase in the volume of metal being deformed per unit cross section area of crack surface. The decrease in toughness at thicknesses greater than t^* is caused by an increase in the area which fractures in low energy, plane strain normal rupture relative to the area which fractures by shear lip (plane stress, shear rupture).

When delamination into thin sheets occurs in the divider orientation the toughness of the divided laminate will be equal to the toughness of the thin sheet of given thickness, t. For maximum toughness in the divider orientation, the base alloy thickness should be chosen at the thickness t^* where K_c is maximum.

Although delamination occurred in the divider orientation of several billets in the experimental program, there was little change in toughness over wrought plate. Figure 14 shows a curve for K_c as a function of thickness for 6Al-4V titanium. Although the data does not extend down to the 0.040 inch thickness used in the present work, K_c for this thickness is well below the maximum value of 160 ksi $\sqrt{\text{in}}$. When delamination occurs in the divider orientation, the composite toughness measured will be the same as for K_c for the thickness of base material used in the composite. For this work, the K_c value for a base metal thickness of 0.040 inch can be found from the fracture toughness data for billets where delamination took place. By using this value as another point on Figure 14 and drawing a representative curve, it is evident that the 0.040 inch material does not exhibit the maximum value of plane stress toughness.

4 Specimen thickness, (B)

In very thin specimens where no triaxiality is obtained (plane stress deformation), no delamination would be expected to take place. Once the specimen is of sufficient thickness such that plane strain conditions are present, there should be no effect of specimen size on the delamination process.

5 Thickness of the interface material, (t_i)

As the thickness of the interface material is decreased, the triaxial constraint in the interface increases. This was shown by West *et al.*[6] for brazed joints under uniaxial loading. When the triaxiality of the interface is increased, the theory of McClintock[5] predicts that the critical strain for fracture, σ_f, will decrease. Thus a lower critical stress σ_f and a lower stress intensity factor will be necessary to initiate delamination (Eq. 12). Therefore, a decrease in interface thickness will increase the possibility of delamination.

This effect has been shown experimentally. For the arrester orientation, the heat treated material containing 0.010 inch titanium foil did not delaminate while the thinner 0.005 inch foil did delaminate and arrest the crack. No attempt to optimize the interface thickness was made during this work.

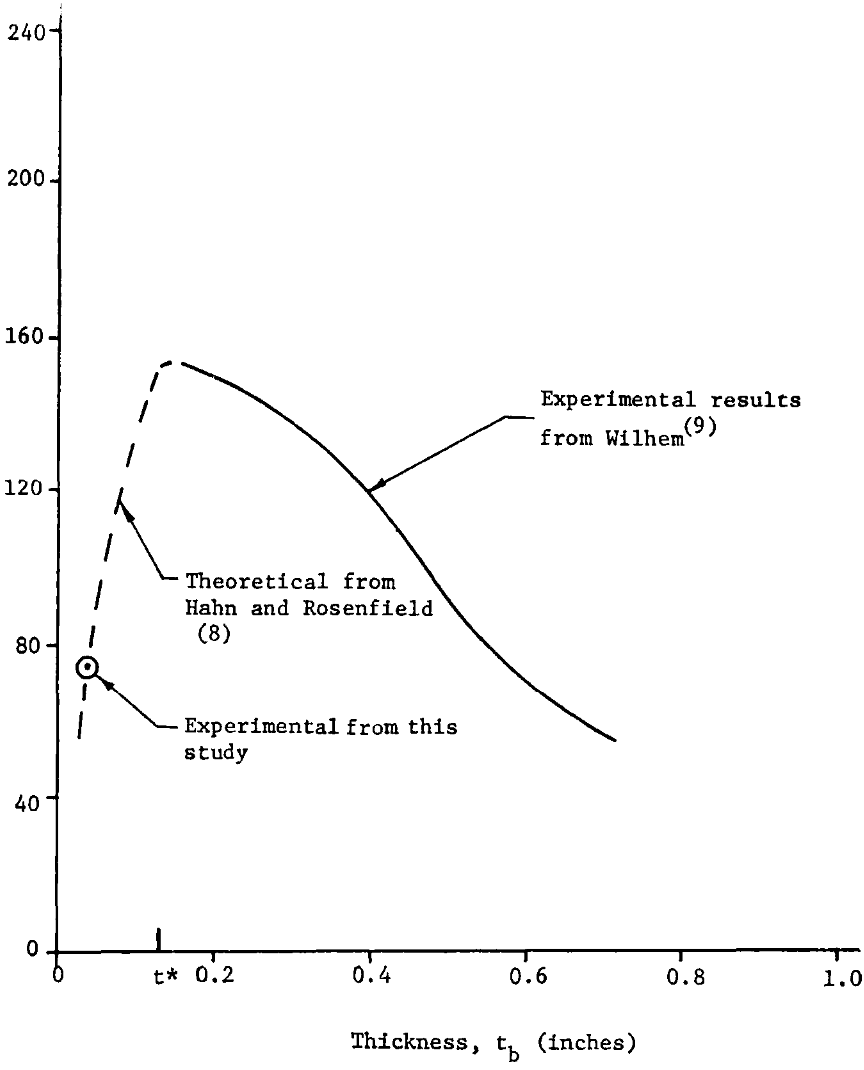


FIGURE 14 K_c vs. thickness for 6Al–4V titanium material.

6 Fracture strength of the interface material, (σ_f)

A change in yield strength of the interface will change the stress-strain relation shown in Figure 13. A decrease in yield strength will generally lower the fracture stress σ_f for a given ductility ε_f. This will reduce the stress intensity factor required for delamination. The effect of variations in interface yield strength or fracture strength was not studied independently.

The combined effect of interface thickness and yield strength can be determined by comparison of the results for Billets No. 3 and No. 6. Whereas, no delamination occurred with the higher yield strength, thicker titanium interleaf, delamination and splitting did take place with the thin, low yield strength aluminum foil interleaf.

A second factor which governs the fracture strength is the ductility of the interface. For the same yield strength, an interface with lower ductility will have a lower fracture strength. Therefore, as the ductility of the interface is decreased, the delamination process is favored. This has been shown experimentally where the microvoid interfaces were used. Delamination and splitting occurred in almost all cases due to the restricted deformation zone of this type of interface. A low ductility adhesive material used as an interface, although not tested experimentally would also show this behavior.

TABLE III

Effect of various parameters on delamination characteristics of laminate composites

Parameter		Significant Effect
Base Alloy Fracture Toughness	$(K_{Ic})_b$	Increase favors delamination
Base Alloy Yield Strength	$(\sigma_Y)_b$	Increase favors delamination
Base Alloy Sheet Thickness	$(t)_b$	Optimum gives maximum K_c
Specimen Thickness	B	None if plane strain conditions prevail
Interface Material Thickness	$(t)_i$	Decrease favors delamination
Interface Material Fracture Strength	σ_f	Decrease favors delamination

The effect each of these six physical and mechanical properties has on the delamination characteristics of the multi-laminate composite is summarized in Table III.

CONCLUSIONS AND RECOMMENDATIONS

The use of composite materials consisting of lamellae of 6Al-4V titanium with a weak interface has been shown to be an effective way of increasing the toughness of high strength Ti-6Al-4V material. Two types of interface were utilized, microvoid sheets and a thin foil. From the reliability standpoint, the foil type of interface with a full strength diffusion bond is superior to the controlled porosity diffusion bond which is weakened by the presence of microvoids. During billet fabrication, the formation of a uniform distribution of microvoids is difficult to achieve and therefore the strength of the interface varies from one area to another.[10] The use of a full strength bond without microvoid formation is much more desirable because there are fewer quality control problems involved in producing the bond itself.

The use of controlled diffusion bonding as an effective means of increasing toughness has been attempted in this study and in past experimentation.

In the arrester geometry the toughness is much higher if delamination occurs and gross yielding follows. In the divider orientation, splitting could increase the fracture toughness to K_c for the sheet thickness used in the composite. This increase in turn leads to an increase in the critical crack size for the material.

A second effect which can occur in laminate composites has a greater consequence. Assume we have a flaw which becomes the initiation site of a crack, which then grows at a certain cyclic stress level. This flaw may be a surface flaw as shown by points A and B in Figure 15 or an internal flaw as

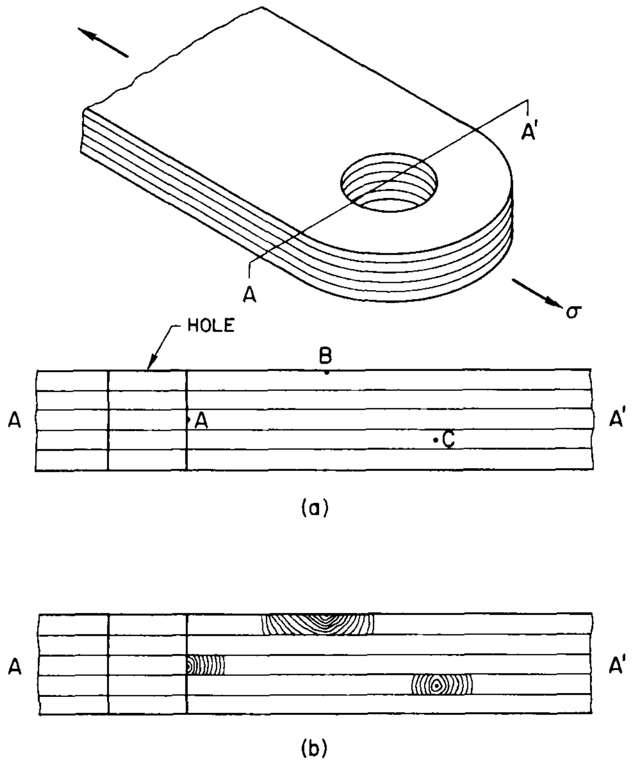


FIGURE 15 Crack channeling in laminate composites.

shown by point C. As these cracks grow toward the interface, delamination at the interface will occur if the stress intensity factor, K_I, is of sufficient magnitude $(K_I)_s$. If delamination takes place then the crack is effectively channeled. As long as the area of the crack is a small fraction of the cross section of the load bearing member, overloading of the remaining area will not take place. The structure is essentially composed of a series of redundant elements. This is of significant value in aerospace structures where redundancy in design is required to meet fail-safe criteria. By use of laminate composite materials the redundancy is built into the material itself, rather than through the use of dual, full size structural elements.

The use of a multi-laminate composite containing weak interfaces has a great potential for practical applications. Optimization of the composite may be possible through consideration of the factors discussed previously. From these parameters, thin interleaf materials with relatively low yield strengths seem desirable. Also for maximum increase of toughness in the divider orientation, the base metal thickness should be chosen so that K_c is maximum. Further work is necessary to determine what materials and thicknesses should be used to obtain the maximum increase in toughness while maintaining a good ratio of transverse to longitudinal strength. While the six parameters discussed will help in the optimization process, other factors must be considered. These include cost, ease of bonding, and other fabrication problems. Further work on the fatigue properties and the stress corrosion characteristics is also necessary before these types of materials may be applied to practical applications.

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