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EFFECT OF ANGLEPLYING AND MATRIX ENHANCEMENT ON IMPACT-RESISTANT BORON/ALUMINUM COMPOSITES

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16. Abstract

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EFFECT OF ANGLEPLYING AND MATRIX ENHANCEMENT ON IMPACT-RESISTANT BORON/ALUMINUM COMPOSITES

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SUMMARY

Efforts to improve the impact resistance of boron/aluminum (B/Al) are reviewed and analyzed. Tensile and dynamic modulus tests, nonstandard thin-sheet Charpy and Izod impact tests, and standard full-size Charpy impact tests were conducted on 0.20-mm-(8-mil-) diameter-B/1100 Al matrix composite specimens. Angleplies investigated ranged from unidirectional to $\pm 30^{\circ}$. Impact failure mode mechanisms of B/Al are proposed, and an attempt is made to exploit these mechanisms to maximize impact resistance.

The best compromise between reduced longitudinal properties and increased transverse properties appeared to be obtained with $\pm 15^{\circ}$ angleply. Angleplies up to $\pm 15^{\circ}$ improved the transverse properties of B/1100 Al without appreciably degrading the longitudinal properties, but higher angleplies were too weak and ductile to allow sufficient energy absorption. Matrix enhancement, with titanium foils, improved the impact strength of brittle composites but reduced the impact strength of ductile, improved impact-resistant B/1100 Al composites.

The pendulum impact strengths of improved B/Al composites were higher than that of notched titanium and appear to be high enough to give sufficient foreign-object-damage protection to warrant consideration of B/Al for application to fan and compressor blades in aircraft gas turbine engines.

INTRODUCTION

The lack of foreign-object-damage (FOD) resistance has been a major obstacle to the use of composites as fan blades in aircraft engines. To overcome some of these problems, the NASA Lewis Research Center is conducting research into impact improvement of both polymer- and metal-matrix composites for blade applications. Results reported in references 1 and 2 show that the impact properties of unidirectional boron/aluminum (B/Al) composites were significantly increased by proper selection of materials and processing. Composites with 8-mil-diameter boron fibers in a ductile 1100 Al matrix had the highest impact strength in thin-sheet and full-size impact tests. Impact energy absorption was increased with this materials combination as a result of increased matrix shear deformation and multiple fiber breakage.

While the longitudinal tensile and impact strengths of these composites were increased to a usable level, the transverse properties were lower than desired. Two methods are commonly used to increase the transverse properties of composites: angleplying, in which the plies are oriented at a specified angle to the loading direction; and matrix enhancement, in which foils or fibers of a third material are incorporated to modify the properties of the matrix.

This report reviews the results of NASA programs to improve the impact resistance of 0.20-mm-(8-mil-) diameter-B/1100 Al composites by angleplying and matrix enhancement. The test program consisted of room-temperature tensile and dynamic modulus-of-elasticity tests, thin-sheet Izod and Charpy impact tests, and full-size Charpy impact tests. Impact failure modes are proposed and are related to the results obtained.

MATERIALS AND PROCEDURE

Specimen Preparation

All B/Al panels for the studies reported herein were produced by Avco Corp. and nominally contained 48 volume percent of 0.20-mm-(8-mil-) diameter boron fibers in an 1100 Al matrix and were made by press diffusion bonding of fiber layups between matrix foils. Panels were bonded in vacuum with $\pm 7^{\circ}$, $\pm 15^{\circ}$, $\pm 22^{\circ}$, and $\pm 30^{\circ}$ angleplies. Data from tests on unidirectional B/Al composites (ref. 1) are included for comparison. Because of the standard nomenclature used in the aerospace industry, the boron fiber diameter is referred to in mils, rather than SI units, throughout the remainder of this report.

Based upon the results of reference 1, standard fabrication conditions of 755 K (900° F) for 0.5 hour at 34 MPa (5 ksi) were used for all panels. Two types of panels were produced. The eight-ply panels used for tensile, dynamic modulus, and thin-sheet impact tests were 30.5 cm x 30.5 cm (12 in. x 12 in.) and 0.20 cm (0.080 in.) thick. The 40-ply panels for full-size Charpy test specimens were 15 cm x 15 cm x 1 cm (6 in. x 6 in. x 0.4 in.). The full-size Charpy specimens were surface ground to ASTM specifications, and a 45° notch was cut into one face.

Specimen Geometry

Because of the anisotropic properties of composites, specimen geometry must be uniquely defined in terms of fiber direction, pressing direction, and notch location. These geometries are shown in figure 1. The LT, TT, and TL geometries are defined in references 3 and 4. The LT geometry is further defined in reference 2 as LT, where the notch is on a side normal to the pressing direction, and LT(s), where the notch is on a side parallel to the pressing direction. Specimens with LT, LT(s), and TT geometries were tested in the studies reported herein.

Dynamic Modulus of Elasticity Tests

The room-temperature dynamic modulus of elasticity of B/Al composites was measured by sonic methods. Rectangular specimens, cut from thin-sheet panels, were 13 cm x 0.95 cm x 0.20 cm (5 in. x 0.375 in. x 0.080 in.). Modulus was determined by measurement of the resonant frequency of the fundamental flexural mode of vibration. Resonance was determined from maximum oscilloscope and voltmeter deflections from stereo phonograph cartridges placed on each end of the specimen. Resonance was verified to be fundamental flexural vibration by traversing the receiving cartridge along the length of the specimen and observing the nodal points and the change of oscilloscope ellipse angle. The dynamic modulus was calculated by using equations from reference 5, based upon reference 6.

Tensile Tests

Room-temperature tensile tests were conducted on an Instron screw-driven-crosshead testing machine. A strain-gage extensometer with a 2.54-cm (1-in.) gage length was used to measure strain on an X-Y recorder. A crosshead speed of

0.025 cm/min (0.01 in/min) was used at the start of the tests to measure initial elastic strain. The crosshead speed was increased to 0.25 cm/min (0.1 in/min) after a test was underway.

Two specimen geometries were used. Longitudinal specimens (testing direction parallel to fiber axis) were 13 cm x 0.95 cm x 0.20 cm (5 in. x 0.375 in. x 0.080in.). Aluminum doublers 4.4 cm (1.75 in.) long were adhesively bonded to the ends of the flat specimens, leaving a gage length of 3 cm (1.25 in.). Transverse specimens (testing direction perpendicular to fiber axis) were 7.6 cm x 0.95 cm (3 in. x 0.375 in.). Doublers for these specimens were slightly under 2.5 cm (1 in.) long.

Impact Tests

Three types of pendulum impact tests were conducted: unnotched thin-sheet Izod, unnotched thin-sheet Charpy, and notched full-size Charpy. All impact tests were conducted at room temperature. The thin-sheet tests were made because they are more economical in terms of material and machining costs and serve as a convenient screening technique. The cantilever mounting of the thin-sheet Izod specimen tends to simulate the behavior of a modern, thin-airfoil fan blade in engine operation. The thin-sheet Charpy tests provided an indication of the behavior of the material without restraint. The full-size Charpy test results provided a comparison of standard specimens with literature values for other materials.

Thin-sheet Izod tests were conducted on a Bell Telephone Laboratories miniature Izod impact testing machine. With appropriate weights on the tup, the capacity was 7.0 J (61.8 in-lb). The calibration and operation of this machine are described in reference 7. Nonstandard unnotched specimens were 3.7 cm x 0.6 cm (1.5 in. x 0.25 in.).

Thin-sheet Charpy tests were conducted on a TMI low-capacity impact testing machine with the grips modified to give standard ASTM separation. The tup used had a maximum capacity of 5.4 J (4 ft-lb). Nonstandard unnotched specimens were 3.7 cm x 0.6 cm (1.5 in. x 0.25 in.).

Full-size Charpy impact tests were conducted on a Rheile impact testing machine with a capacity of 163 J (120 ft-lb). Tests and specimens were made according to ASTM standard E23-66 (ref. 8). Some of the specimens were slightly undersize in thickness. In these cases, the notch depth was slightly reduced so that the remaining material under the notch met the ASTM standard.

RESULTS

Dynamic Modulus-of-Elasticity Tests

Room-temperature dynamic modulus tests were conducted in the longitudinal and transverse directions, and the results of these tests are presented in table I and figure 2. The longitudinal modulus decreased smoothly with increasing angleply. The transverse modulus remained constant up to $\pm 22^{\circ}$ angleply, after which it dropped.

Tensile Tests

Ultimate tensile strengths of 8-mil-diameter-B/1100 Al matrix angleply composites are presented in table II. Longitudinal tensile strength decreased with increasing angleply, as shown in figure 3. Stress-strain curves are shown in figure 4. Figure 4(a) shows stress-strain curves for specimens tested in the longitudinal direction. These curves were plotted until the load started dropping, as indicated by the arrows. The rate of load reduction beyond the strain indicated by the arrows became progressively more gradual with increasing angleply. Unidirectional specimens showed linear behavior to failure. With increasing angleply, nonlinearity and strain to maximum load increased. At failure, the specimens separated along the axes of the angleplied fiber layers. Failed longitudinal tensile test specimens (fig. 5(a)) show that this additional strain was localized and was not indicative of the overall specimen behavior.

The transverse tensile strength increased slightly with increasing angleply. The stress-strain curves in figure 4(b) show that angleplying increased the strain to failure. Failed transverse tensile test specimens (fig. 5(b)) also show separation along the angleply axes.

Impact Tests

Thin-sheet Izod impact test results are presented in table III, thin-sheet Charpy results in table IV, and full-size Charpy results in table V. The unidirectional full-size Charpy specimens made for the in-house program were defective and failed by delamination at energies below those of properly bonded composites (ref. 1). Results from these specimens are included in table V but are not plotted in the figures. Data from reference 2 were used for the unidirectional results.

Results from these tests are plotted in figure 6. The area-compensated impact strength of full-size Charpy specimens was much higher than that of thin-sheet specimens. Failed specimens are shown in figure 7 for thin-sheet Izod tests, in figure 8 for thin-sheet Charpy tests, and in figure 9 for full-size Charpy tests.

Two types of failure behavior were observed in the angleply B/1100 Al specimens. In properly bonded full-size Charpy specimens and in unidirectional and low-angleply thin-sheet specimens, failure occurred by fracture of all the fibers in the cross section and with matrix shear prior to fiber failure. This type of behavior was considered normal failure.

Higher angleply thin-sheet specimens underwent bending distortion but were pushed through the grips at low impact energies with a minimum of fiber breakage. These specimens were considered not to have fractured but to have been extruded through the grips. Angleplied full-size Charpy specimens failed in a normal manner. Unidirectional thin-sheet specimens failed with less plastic deformation than did angleply thin-sheet specimens.

DISCUSSION

Some of the factors influencing toughness of B/Al composites and the relation of failure mode to impact energy absorption of unidirectional composites are discussed in reference 1. The following discussion relates the effect of angleplying and matrix enhancement to the impact properties of B/Al composites.

Effect of Angleplying

Because of the anisotropic nature of composites, the transverse properties of unidirectional composites may not be high enough to withstand stresses encountered during component service. Angleplying can be used to improve transverse properties, but it is accomplished with a corresponding penalty in longitudinal properties.

The results of studies with 8-mil-diameter-B/1100 Al composites having angleplies of $\pm 7^{\circ}$, $\pm 15^{\circ}$, $\pm 22^{\circ}$, and $\pm 30^{\circ}$ are reported herein. In addition, results from tensile and full-size Charpy impact tests on 1100 and 5052 Al matrix composites with three different angleply layups are reported in reference 2. The first angleply consisted of 50 percent unidirectional plies in the central core with 25 percent alternating $\pm 45^{\circ}$ shells on each side of the core. The second angleply consisted of repetitive units of 0°, $+22^{\circ}$, and -22° plies. The third consisted of alternating $\pm 15^{\circ}$ plies.

The $\pm 45^{\circ}$ shell - 0° core configuration had the best combination of transverse tensile and impact strengths but also had the lowest LT strengths. The $\pm 22^{\circ}$,0° angleply gave slightly lower TT impact and tensile strength results than the $\pm 15^{\circ}$ angleply. This was probably due to the 0° fibers, which would give adverse results in the transverse direction. The results of full-size Charpy tests on angleply specimens from reference 2 are summarized in figure 10. The best combination of longitudinal and transverse impact and tensile results was obtained with the $\pm 15^{\circ}$ angleply.

Reference 9 presents theoretical stress distributions from impact shock wave propagation through composites of various angleply configurations. These calculations showed that maximum stress levels occurred immediately after impact and propagated along the angleply axes. The best angleply for resisting flexural impact stresses was calculated to be $\pm 15^{\circ}$. At $\pm 15^{\circ}$, the stress level was 34 percent lower than at $\pm 45^{\circ}$. There did not appear to be a significant difference in the interlaminar shear stresses with layup angle, although there were differences in stress distributions. These studies were based on resin matrix composites and considered only high-velocity impact conditions similar to ballistic impact. However, the findings are probably also applicable to B/Al composites under lower velocity pendulum impact conditions. These calculations predict the results obtained in reference 2, where a $\pm 15^{\circ}$ angleply was found to give the best combination of longitudinal and transverse tensile and impact results.

Longitudinal tensile strength of B/1100 Al decreased with increasing angleply. This strength reduction was associated with a decrease in elastic strain and an increase in the nonlinear strain range, as shown in the stress-strain curves of figure 4. Contrary to expected results, the transverse tensile strength was not increased appreciably with angleply. Likewise the transverse modulus of elasticity remained constant up to $\pm 22^{\circ}$ angleply, after which it decreased slightly. Results for B/1100 Al and B/5052 Al composites, reported in reference 2, followed the expected behavior of increasing transverse tensile strength with increased angleply. The longitudinal and transverse tensile strength results obtained in the investigation reported herein were lower than those reported in reference 2. These differences were attributable to the differences in bond strength caused by the different bonding temperatures used in the two studies.

When tested in the longitudinal direction, unidirectional full-size Charpy and thin-sheet specimens had higher tensile and impact strengths than any of the angleplies (figs. 3 and 6). The $\pm 7^{\circ}$ angleply had a minor loss in impact and tensile strength as compared with unidirectional composites. The maximum angleply that allowed the fiber properties to be utilized was $\pm 15^{\circ}$, where the fibers fractured after attaining sufficient strain to give high stresses and impact energies. At angleplies greater than $\pm 15^{\circ}$, the low strength and nonlinear stress-strain behavior allowed the composites to strain without applying sufficient stress to the fibers for them to make their maximum contribution to the properties of the composite.

Two types of failure were observed for angleply B/Al composites. One in which the specimen failed by fiber fracture upon impact, and one in which the specimen was extruded through the grips without fiber fracture. Increasing angleply decreased the impact strength for fiber-fracture-type failures. Full-size Charpy specimens with $\pm 22^{\circ}$ and $\pm 30^{\circ}$ angleplies failed in a normal manner but with less shear displacement than specimens with lower angleplies (fig. 9). Thin-sheet angleply specimens that underwent fiber failure failed in a manner similar to that of the full-size Charpy specimens. Thin-sheet angleply specimens that failed by extrusion without fiber fracture, $\pm 22^{\circ}$ and $\pm 30^{\circ}$ specimens, underwent considerable stretching and distortion during impact testing. These specimens also underwent a large amount of shear, and some surface striation was observed along the angleply axes (figs. 7 and 8).

Differences in the results from full-size Charpy and thin-sheet impact tests are related to specimen thickness and the failure mechanisms that are involved. References 10 to 12 report a transition in fracture and delamination behavior at a thickness of 0.25 cm (0.1 in.). Below this thickness, plane stress conditions existed and delamination stresses were very high. Fiber/matrix bond failure occurred because of shear stress concentration at the notch tip. Above a thickness of 0.25 cm (0.1 in.), plane strain conditions existed. The transverse tensile stresses at the notch tip caused fiber/matrix bond failure at lower stresses, and the stress to cause delamination remained constant. In both cases, after the notched section delaminated, the remaining section was notch-insensitive and failed as if a notch had not been present (ref. 10).

Angleplying changes the behavior from that described in references 10 to 12 for unidirectional specimens. Angleplying changes the stress distribution within the composite and eliminates planes along which delamination can readily occur. The low tensile strengths of high-angleply composites allow deformation at low flow stresses. Thus, the stress required to cause delamination may not be reached.

Full-size Charpy and nonstandard thin-sheet impact test results showed similar trends for specimens where failure occurred by fiber fracture upon impact. This indicates that thin-sheet impact tests can be used as a screening tool to rank impact behavior of various B/Al composites where failure occurred. Extrapolation of impact strength values from one type of test to another cannot be made accurately. Also rankings are invalid for specimens undergoing extrusion without failure, such as the apparent increase in impact strength of the $\pm 22^{\circ}$ and $\pm 30^{\circ}$ angleply specimens in the thin-sheet Charpy results shown in figure 6. The indicated impact strengths were actually a measure of either (1) impact strength, if the material was strong enough or brittle enough not to deform excessively and fracture occurred; (2) bending stress, if the material was ductile enough to allow it to be extruded through the holders without fracture; or (3) a combination of the two, where the material partially deformed and partially failed.

Effect of Matrix Enhancement

Another possible method of improving the transverse strength of B/Al composites is matrix enhancement, where a third material, either in foil or fiber form, is placed between the aluminum-matrix foils to modify the matrix properties.

References 3 and 4 report that the addition of 6-volume-percent stainless steel wire, oriented in the transverse direction, increased the TT-specimen Charpy impact strength of 4.2-mil-diameter-Borsic/6061 Al composites from 1.5 J to 6 J (1.1 ft-lb to 4.5 ft-lb). In later work (ref. 13), LT-specimen Charpy impact strength of 4-mil-diameter-B/6061 Al was increased by 60 percent, to 26 J (19 ft-lb), with an accompanying increase in transverse tensile strength, by using a dual alloy matrix of 6061/1100 Al. It was suggested that impact strength and transverse properties could be increased further by using titanium-foil matrix enhancement.

To determine the effect of matrix enhancement, foils of Ti-6Al-4V were used in 5.6- and 8-mil-diameter-B/1100 Al composites in reference 2. Results showed that matrix enhancement reduced the longitudinal tensile strength by about 15 percent and reduced the full-size LT-specimen Charpy impact strength by over 50 percent for both composites. The transverse tensile strength was increased from 65 MPa to 266 MPa (10 ksi to 39 ksi). However, the TT-specimen Charpy impact strength was only increased from 1.4 J to 4.1 J (1 ft-lb to 3 ft-lb). This slight increase in transverse impact strength would not justify the sacrifice in longitudinal impact. The same trends held for angleply B/1100 Al composites. The LT-specimen impact strength was reduced by more than 50 percent, but the TT-specimen impact strength was virtually unchanged by titanium-foil enhancement. Bar graphs in figure 11 show the effect of matrix enhancement on full-size Charpy impact strength of unidirectional and $\pm 22^{\circ}, 0^{\circ}$ angleply composites for the three specimen geometries (LT, LT(s), and TT) tested.

The data obtained in reference 2 seem to differ from other reported results. It is generally thought that titanium-foil interleaves should increase the impact strength of B/Al. Titanium is very impact resistant in the unnotched condition. However, a notch reduces impact strength from 318 J to 23 J (220 ft-lb to 15 ft-lb). Thus in a notched impact test, titanium-foil matrix enhancement would only improve the impact strength of composites that have impact strengths below that of notched titanium itself. Titanium foil will restrain the matrix from flowing, making fracture and crack initiation more difficult and thereby increasing the impact strength of brittle composites. It also provides delamination planes for low-impact composites, which rely on energy dissipation by delamination to improve impact behavior.

For ductile, impact-resistant B/Al composites, however, the effect of titanium-foil interleaves is different. The matrix ductility restraint imposed by the titanium foil will embrittle more-ductile composites, such as B/1100 Al. By not allowing the matrix to shear, this restraint will not permit the fibers to attain their full strength contribution, thus reducing the plastic shear contribution from a ductile Al matrix. Fractographs (fig. 12) compare the more brittle planar fracture surface of specimens with matrix enhancement with the jagged surface (bonded fiber/matrix zones projecting out of the fracture surface) of specimens without matrix enhancement.

Reference 2 reports a directionality effect in high-impact B/Al composites without matrix enhancement, where the impact strengths in the LT(s) geometry were much lower than those in the LT geometry. This behavior is explained in reference 1 as being caused by the change in crack propagation from sequential cracking of each ply to simultaneous cracking of all plies. Simultaneous cracking reduced the energy absorption by reducing matrix shear.

Impact strength reductions in B/Al composites with titanium-foil matrix enhancement are also reported in reference 2. The LT(s)-specimen impact strength was reduced much more than the LT-specimen impact strength. This was caused by the planar nature of the foil. This behavior is analogous to that reported in reference 14 for Charpy impact behavior of soft-soldered steel laminates. Two orientations were used (fig. 13), corresponding to LT and LT(s) specimens. The LT geometry was termed a "crack-arrester" laminate. In this case, triaxial stress caused brittle fracture of the first layer, after which the crack was blunted by the interface. Subsequent failure was caused by uniaxial tension, which is unfavorable to brittle cleavage. If delamination did not occur, the cleavage crack in the first layer could be arrested by plastic deformation in the laminate bond. The LT(s) case was termed a "crack-divider" laminate. In this case, the interfaces divide the crack into a series of cracks propagating through the individual subunits. If these are sufficiently thin, the triaxial tension will be relaxed to a state of biaxial tension in each layer, thus inhibiting brittle cleavage fracture. If delamination does not occur, the effect is small unless a large amount of energy is absorbed in fracturing the ductile material.

In LT testing of B/Al composites with titanium-foil matrix enhancement, the notch extends into the flat surface of the foil. As a crack propagates through the specimen, it must either penetrate the foil sheet or cause the titanium-foil/aluminum-matrix interface to delaminate. If the bonding is sufficient, each successive foil will act as a notched titanium sheet and will fail sequentially in a somewhat ductile manner. When tested in the LT(s) direction, the foils no longer fracture sequentially, but instead fracture simultaneously as a unit, in a more brittle manner.

Application of Improved Impact Technology to Aircraft

Gas Turbine Engine Fan Blades

The large increases in pendulum impact strength for improved B/Al composites presented in this report and in references 1 and 2 are very encouraging. The advantages of the improved B/Al composites are shown in figure 14, which compares values from reference 2 with impact strengths of notched titanium and of B/Al produced without the improved-impact-resistance technology. These results provide a basis for expecting that a significant improvement in fan blade performance might be obtained by using improved B/Al composites. However, the results of low-velocity pendulum impact tests on laboratory specimens do not necessarily mean satisfactory foreign-object-damage resistance for complicated fan blade geometries at high-velocity fan blade operating conditions.

Results on blade-like specimens of $\pm 15^{\circ}$-angleply 8-mil-diameter-B/1100 Al are reported in reference 2. These blade-like specimens had a flat, untwisted airfoil-like section and a splayed three-wedge root. The root was placed in a clamp and the specimens were subjected to low-velocity impact tests. Specimens of 8-mil-diameter-B/1100 Al failed at the root-airfoil fillet after considerable shear. Thus the matrix shear displacement probably took place in a manner similar to that for Charpy/Izod standard and thin-sheet specimens. A limited number of high-velocity tests with blade-like specimens were also performed. Figure 15 shows a B/Al blade-shaped specimen after high-velocity ballistic impact with simulated rubber birds (ref. 2). Specimens were able to withstand impact energies up to 250 J (184 ft-lb), the maximum energy tested. The specimens deformed by shear, with deformation primarily in the root area. No delamination was observed, and leached-out fibers indicated no evidence of fiber breakage at the root.

The results of both low-velocity pendulum and high-velocity ballistic impact tests on improved impact-resistant B/Al indicate that successful application of B/Al to fan blades is possible. However, extensive additional studies are required, including single-blade static FOD tests, whirling arm tests, and full-stage engine ground tests. Flight experience must then be accumulated to develop confidence that B/Al is ready for application to engine fan blades. The work described herein represents a start toward this goal. These promising results should serve to further the development of B/Al composites so as to obtain the large payoffs in performance, fuel economy, and cost and weight reduction that are possible with composite fan blades in aircraft gas turbine engines.

SUMMARY OF RESULTS AND CONCLUSIONS

The following major results and conclusions were obtained from NASA programs to improve the impact properties of diffusion-bonded boron/aluminum (B/Al) composites:

1. Pendulum impact test strengths of improved B/Al were higher than that of notched titanium and appear to be high enough to give sufficient foreign-object-damage protection to warrant their consideration for fan and compressor blades in aircraft turbine engines. However, extensive additional studies, such as single-blade static tests, whirling arm tests, and ground engine tests, are required to develop confidence that B/Al composites are ready for such applications.

2. Transverse tensile and impact strengths were increased through the use of angleply fibers. The optimum angleply for the best combination of longitudinal and transverse tensile strength and impact resistance appeared to be about $\pm 15^{\circ}$.

3. Matrix enhancement, by means of titanium-foil interleaves, reduced the longitudinal impact strength of ductile, high-impact-strength B/Al composites.

Lewis Research Center,
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Cleveland, Ohio, May 4, 1976,
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TABLE I. - RESULTS OF DYNAMIC MODULUS-OF-ELASTICITY TESTS

[Fiber, 0.20-mm-(8-mil-) diameter boron; matrix, 1100 aluminum; tested at room temperature.]

Angleply, deg	Dynamic modulus of elasticity			
	Longitudinal		Transverse	
	GPa	Mpsi	GPa	Mpsi
0	218	31.6	130	18.8
	219	31.8	131	19.0
±7	213	30.9	134	19.4
	215	31.1	134	19.4
±15	196	28.4	130	18.8
	196	28.4	128	18.5
±22	186	26.9	125	18.2
	187	27.1	123	17.8
±30	155	22.5	107	15.6
	152	22.1	109	15.8
	155	22.4	109	15.8
	159	23.0	107	15.6

TABLE II. - RESULTS OF TENSILE TESTS

[Fiber, 0.20-mm-(8-mil-) diameter boron; matrix, 1100 aluminum; tested at room temperature.]

Angleply, deg	Ultimate tensile strength			
	Longitudinal		Transverse	
	MPa	ksi	MPa	ksi
0	1175	170.4	32	4.7
	1296	188.0	--	---
±7	1246	180.7	50	7.3
	1201	174.2	--	---
	1109	160.8	--	---
±15	889	128.9	38	5.5
	816	118.3	--	---
±22	416	60.4	37	5.3
	377	54.7	--	---
±30	238	34.5	42	6.1
	206	29.9	--	---

TABLE III. - RESULTS OF THIN-SHEET IZOD IMPACT TESTS

[Fiber, 0.20-mm-(8-mil-) diameter boron; matrix, 1100 aluminum; tested at room temperature.]

Angleply, deg	Longitudinal				Transverse			
	Impact energy		Area-compensated impact strength		Impact energy		Area-compensated impact strength	
	J	ft-lb	kJ/m^2	$ft\text{-}lb/in^2$	J	ft-lb	kJ/m^2	$ft\text{-}lb/in^2$
0	3.28	2.42	233	106.1	0.19	0.14	13	6.0
	2.89	2.13	210	100.1	----	----	--	---
	2.28	1.68	166	78.9	----	----	--	---
	2.25	1.66	169	80.6	----	----	--	---
	>3.00	>2.21	>216	>102.8	----	----	--	---
±7	3.84	2.83	245	116.6	0.27	0.20	17	8.3
	2.90	2.14	209	99.6	----	----	--	---
	2.51	1.85	180	85.2	----	----	--	---
	2.87	2.12	207	98.5	----	----	--	---
±15	2.62	1.93	197	93.6	0.41	0.30	28	13.3
	2.48	1.83	180	85.6	----	----	--	----
±22	1.75	1.29	126	60.1	0.41	0.30	28	13.3
	2.26	1.67	154	73.3	----	----	--	----
±30	1.37	1.01	105	49.9	0.33	0.24	23	11.1
	1.46	1.08	105	50.2	.20	.15	16	7.5
	1.40	1.03	99	47.2	.31	.23	21	10.1
	1.07	.79	76	36.4	----	----	--	----
	1.18	.87	90	42.6	----	----	--	----
	1.46	1.08	104	49.5	----	----	--	----

TABLE IV. - RESULTS OF THIN-SHEET CHARPY IMPACT TESTS

[Fiber, 0.20-mm-(8-mil-) diameter boron; matrix, 1100 aluminum; tested at room temperature.]

Angleply, deg	Longitudinal				Transverse			
	Impact energy		Area-compensated impact strength		Impact energy		Area-compensated impact strength	
	J	ft-lb	kJ/m^2	$ft\text{-}lb/in^2$	J	ft-lb	kJ/m^2	$ft\text{-}lb/in^2$
0	6.13	4.52	327	155.7	0.53	0.39	27	12.8
	4.99	3.68	256	121.7	----	----	--	----
	3.89	2.87	193	91.9	----	----	--	----
	4.01	2.96	198	94.2	----	----	--	----
±7	5.45	4.02	279	132.7	0.41	0.30	22	10.5
	>2.17	>1.59	>108	>51.5	----	----	--	----
	6.79	5.01	332	157.0	----	----	--	----
	5.63	4.15	276	131.2	----	----	--	----
±15	2.93	2.16	166	76.0	0.58	0.43	32	15.1
	3.15	2.32	167	79.4	----	----	--	----
	3.46	2.55	175	83.1	----	----	--	----
	3.62	2.67	183	87.0	----	----	--	----
±22	4.39	3.24	229	109.0	0.95	0.70	51	24.5
	6.30	4.65	319	151.9	----	----	--	----
±30	5.92	4.37	196	140.0	1.32	0.97	70	33.5
	6.70	4.94	338	160.8	1.78	1.31	93	44.1
	6.68	4.93	364	173.4	----	----	--	----

TABLE V. - RESULTS OF FULL-SIZE CHARPY IMPACT TESTS

[Fiber, 0.20-mm-(8-mil-) diameter boron; matrix, 1100 aluminum; tested at room temperature.]

Angleply, deg	Type		Impact strength		Area-compensated impact strength	
	Geometry	Source	J	ft-lb	kJ/m^2	$ft\text{-}lb/in^2$
0	LT	Lewis	[a]24.4	[a]18.0	[a]302	[a]144
			[a]25.1	[a]18.5	[a]311	[a]148
			[a]17.6	[a]13.0	[a]218	[a]104
	LT	Ref. 2	96.3	71.0	1193	568
			91.6	67.5	1134	540
	LT(s)	Lewis	35.9	26.5	445	212
	LT(s)	Ref. 2	47.5	35.0	587	280
±7	LT	Lewis	62.4	46.0	772	368
±15	LT	Lewis	46.8	34.5	579	276
			[a]26.4	[a]19.5	[a]327	[a]156
±22	LT	Lewis	39.3	29.0	487	232
±30	LT	Lewis	28.5	21.0	352	168

[a]Poorly bonded specimen that delaminated during testing; data not plotted in fig. 6.

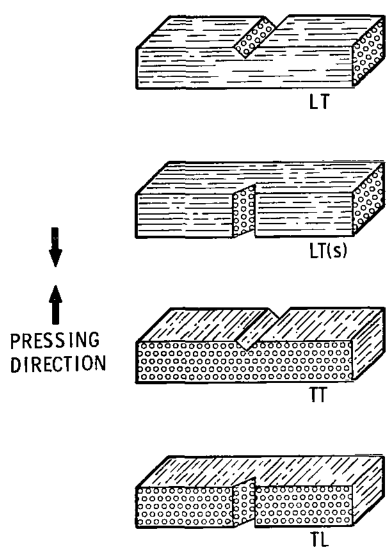


Figure 1. - Charpy impact test specimen geometries.

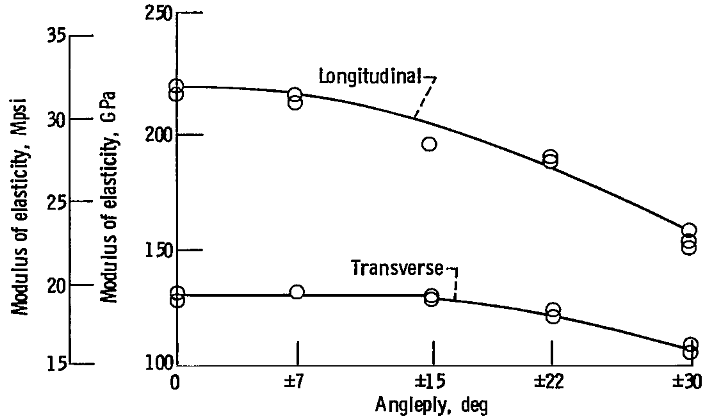


Figure 2. - Room-temperature dynamic modulus of elasticity of 8-mil-diameter-B/1100 Al composites at various angleplies.

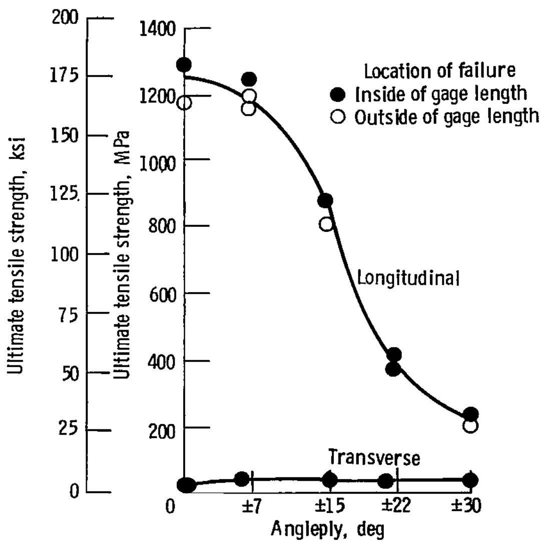
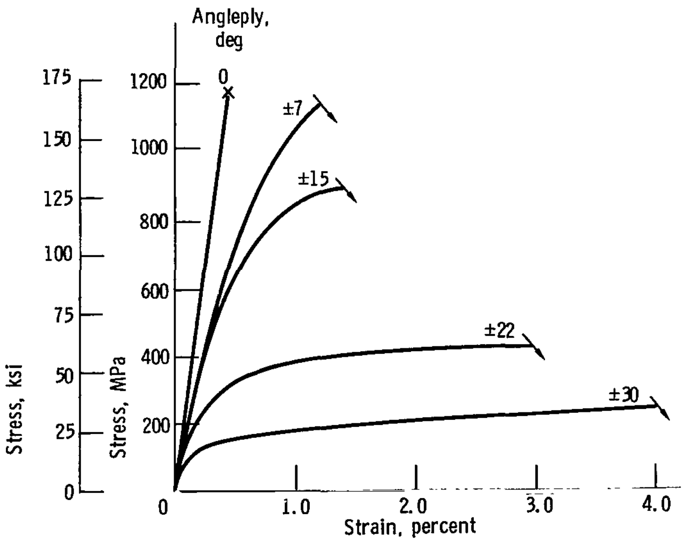
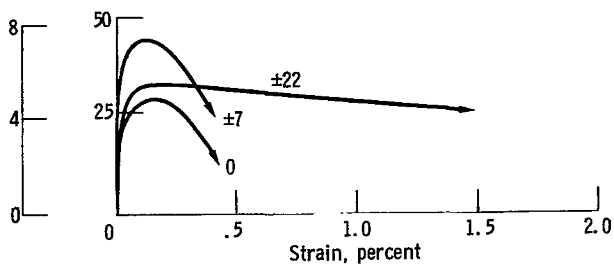


Figure 3. - Room-temperature tensile strength of 8-mil-diameter-B/1100 Al composites.

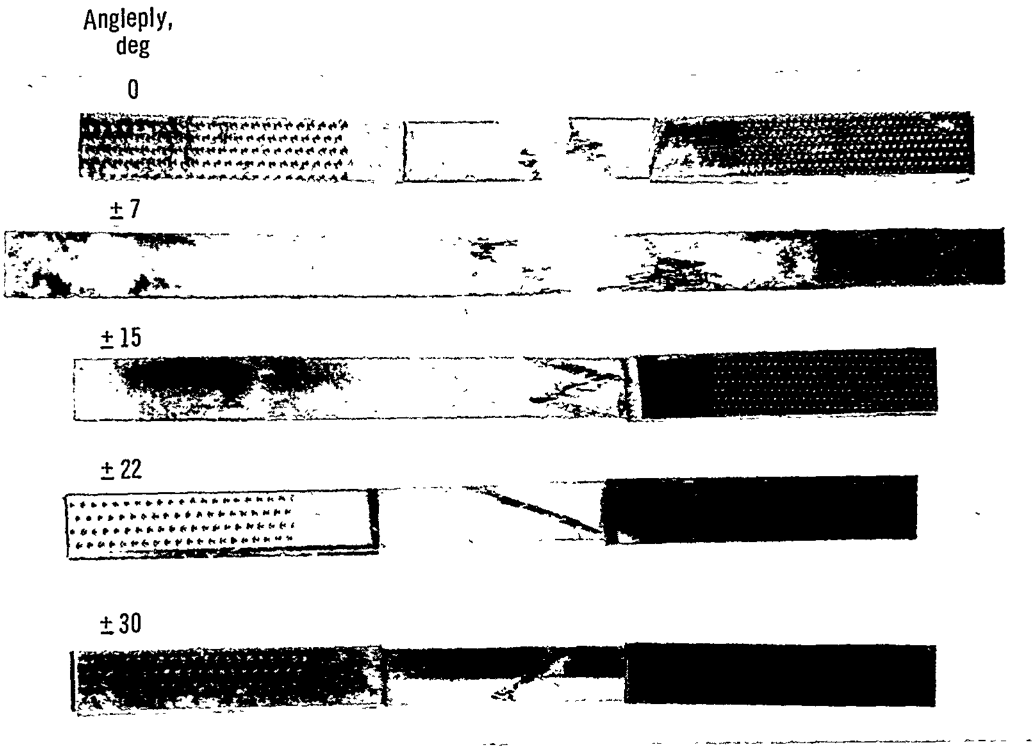


(a) Tested in longitudinal direction.

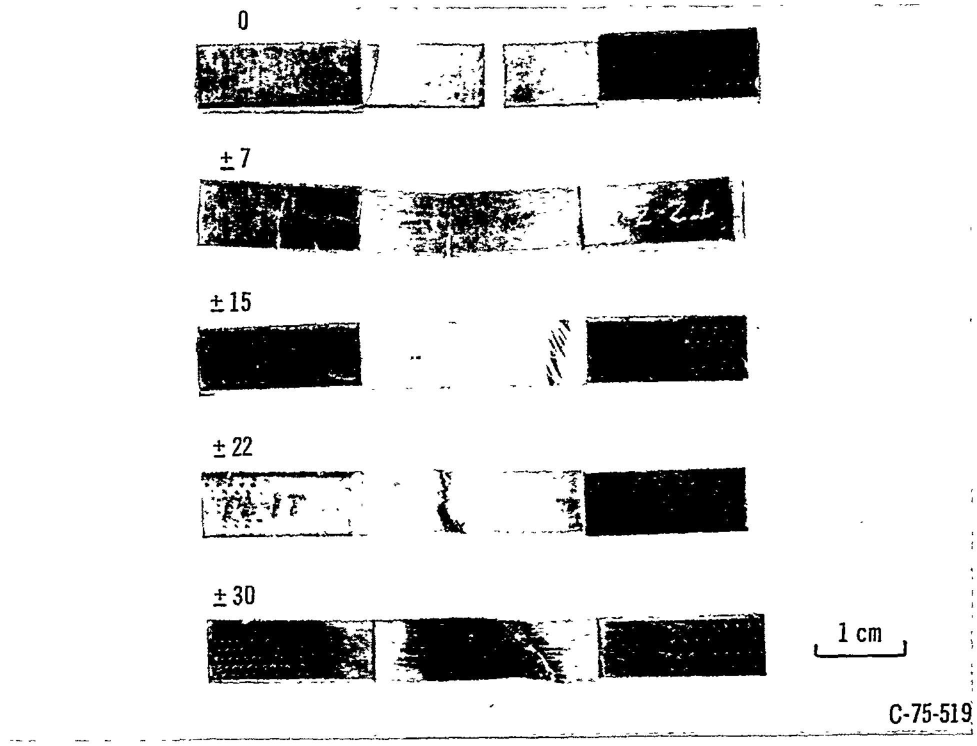


(b) Tested in transverse direction.

Figure 4. - Stress-strain curves of 8-mil-diameter-B/1100 Al composites.



(a) Longitudinal.



(b) Transverse.

Figure 5. - Failed 8-mil-diameter-B/1100 Al tensile test specimens.

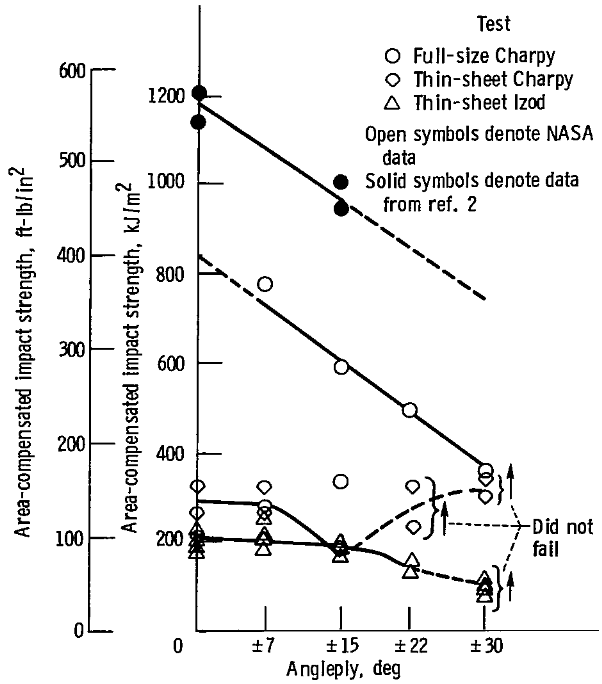
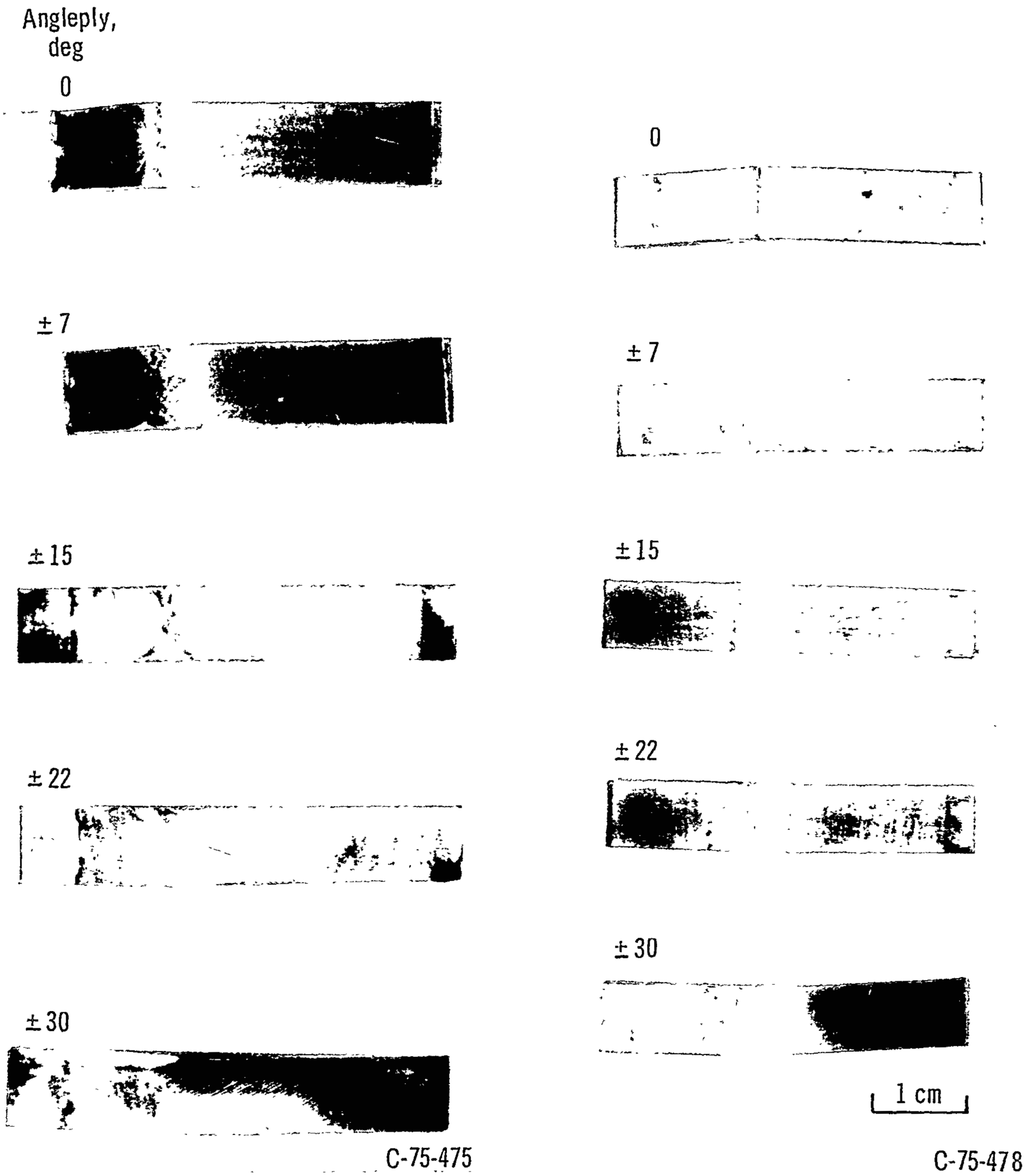


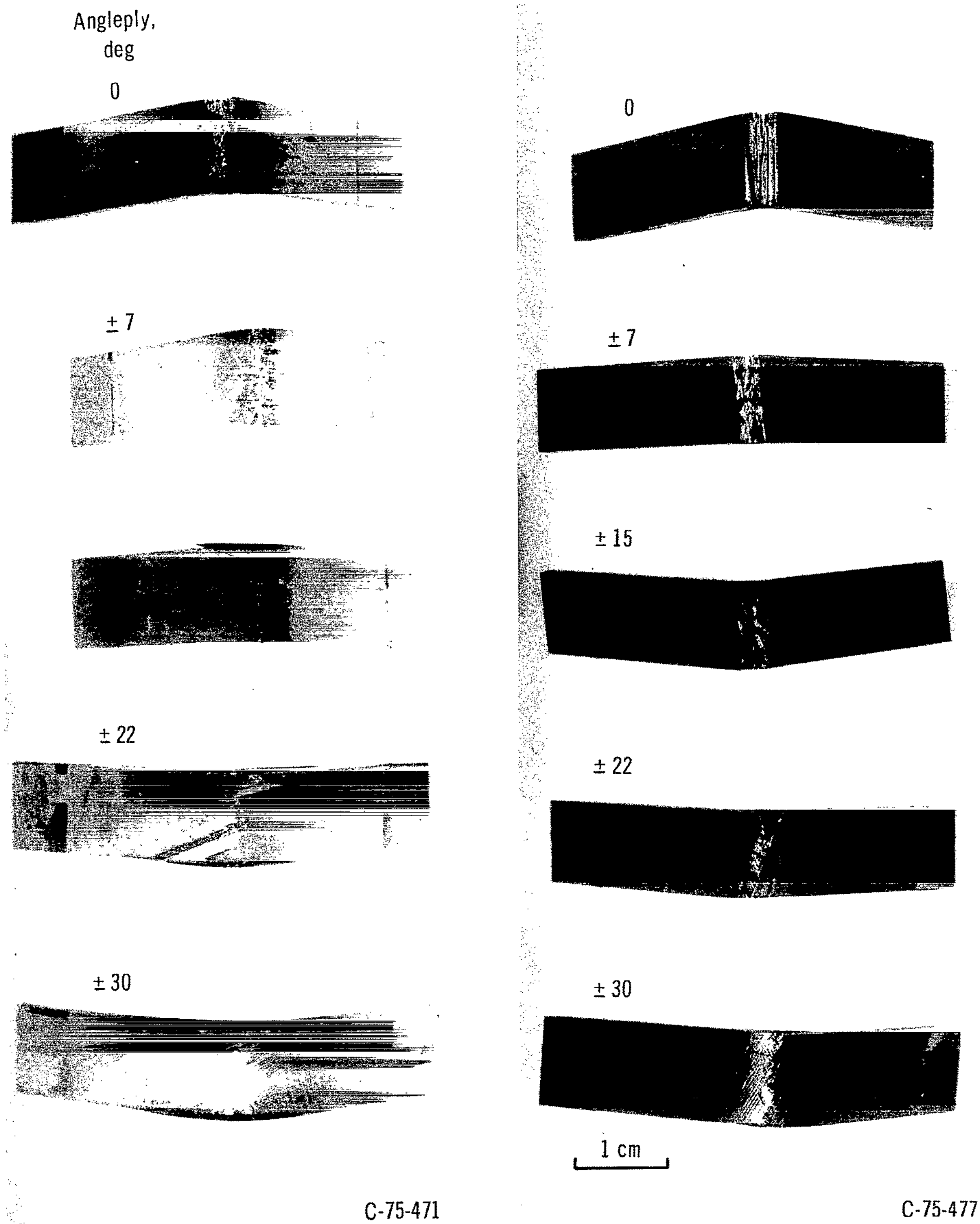
Figure 6. - Area-compensated longitudinal impact strengths of 8-mil-diameter-B/1100 Al composites.



(a) Longitudinal.

(b) Transverse.

Figure 7. - Failed 8-mil-diameter-B/1100 Al thin-sheet Izod impact test specimens.



(a) Longitudinal. (b) Transverse.

Figure 8. - Failed 8-mil-diameter-B/1100 Al thin-sheet Charpy impact test specimens.

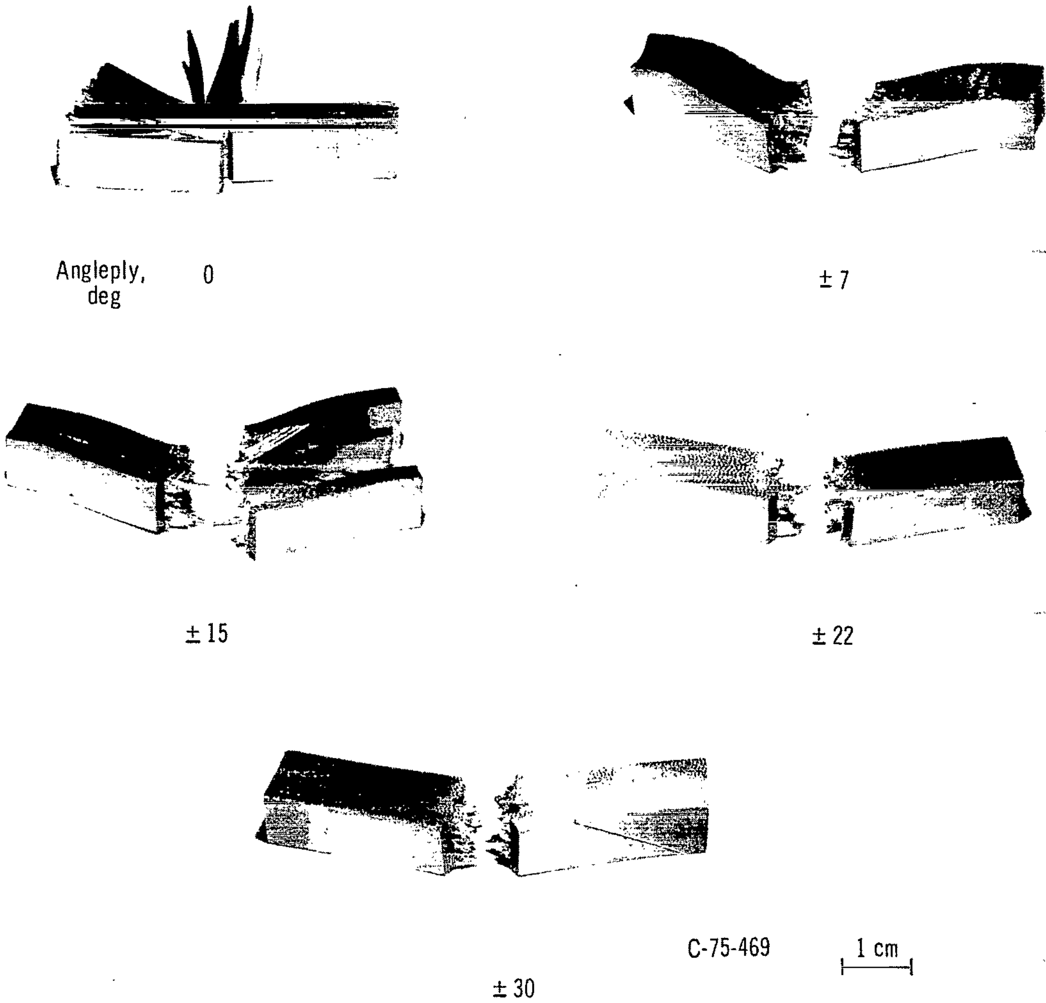


Figure 9. - Failed 8-mil-diameter-B/1100 Al full-size Charpy impact test specimens.

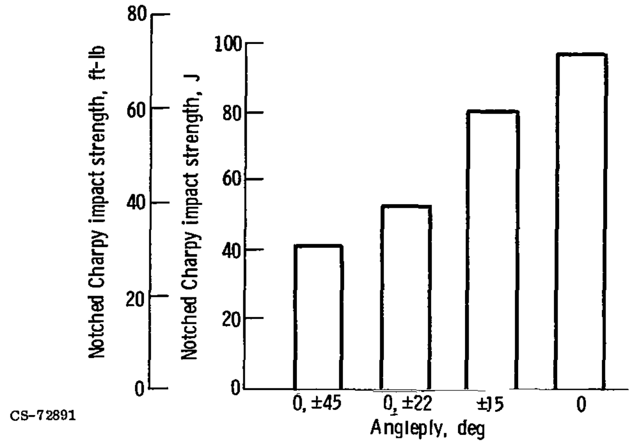


Figure 10. - Summary of results from full-size Charpy impact tests of reference 2.

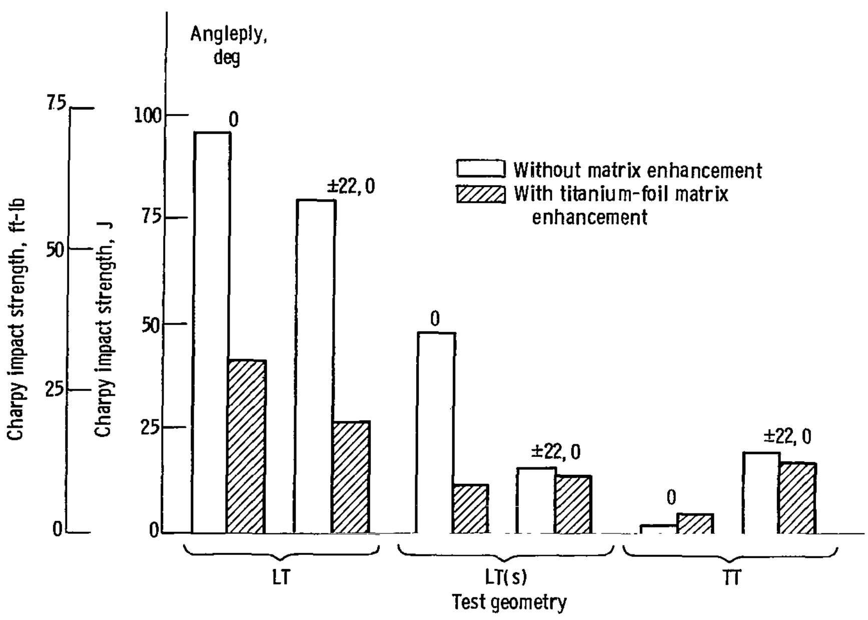
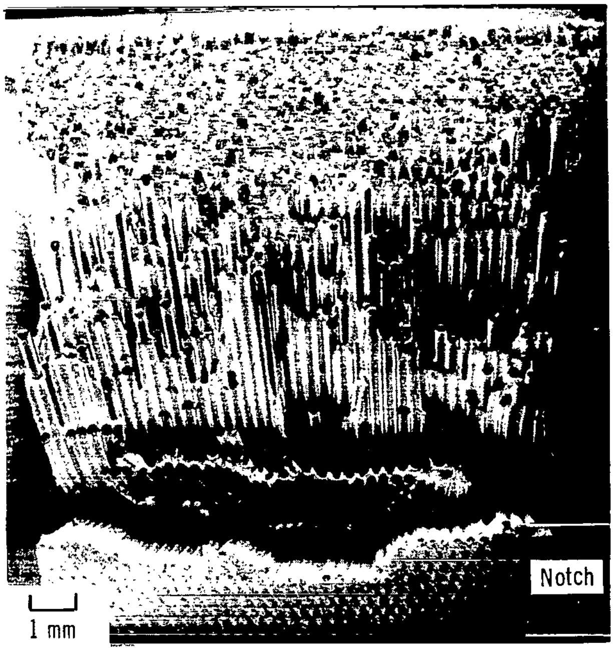
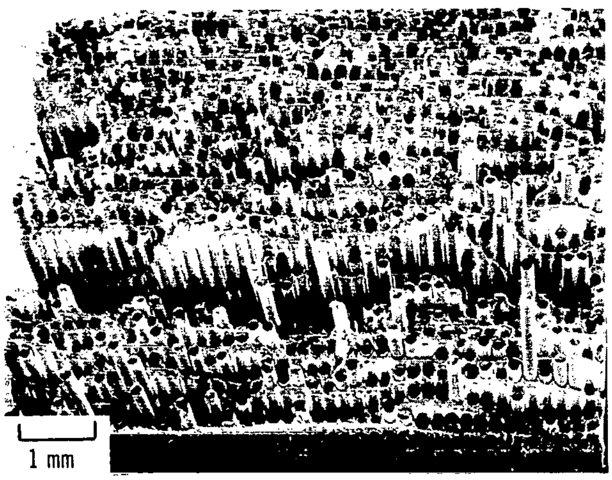


Figure 11. - Effect of titanium-foil matrix enhancement on Charpy impact strength of 8-mil-diameter-B/1100 Al composites.

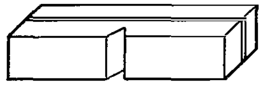


(a) Without matrix enhancement.



(b) With titanium-foil matrix enhancement.

Figure 12. - Comparison of fracture surfaces of 8-mil-diameter-B/1100 Al with and without titanium-foil matrix enhancement. (From ref. 2.)



(a) Crack-arrester laminate.

(b) Crack-divider laminate.

Figure 13. - Orientation of soft-soldered steel laminates. (From ref. 14.)

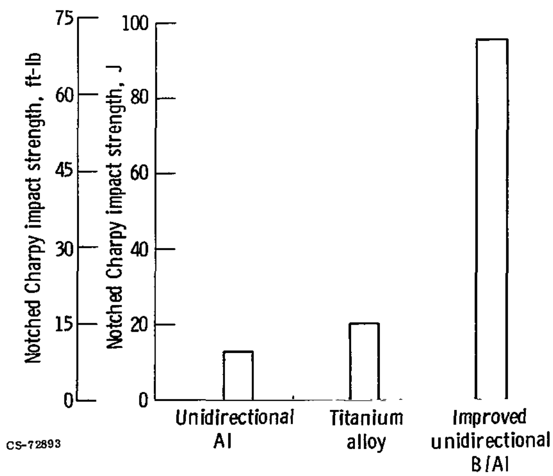


Figure 14. - Improvement of boron/aluminum impact strength.

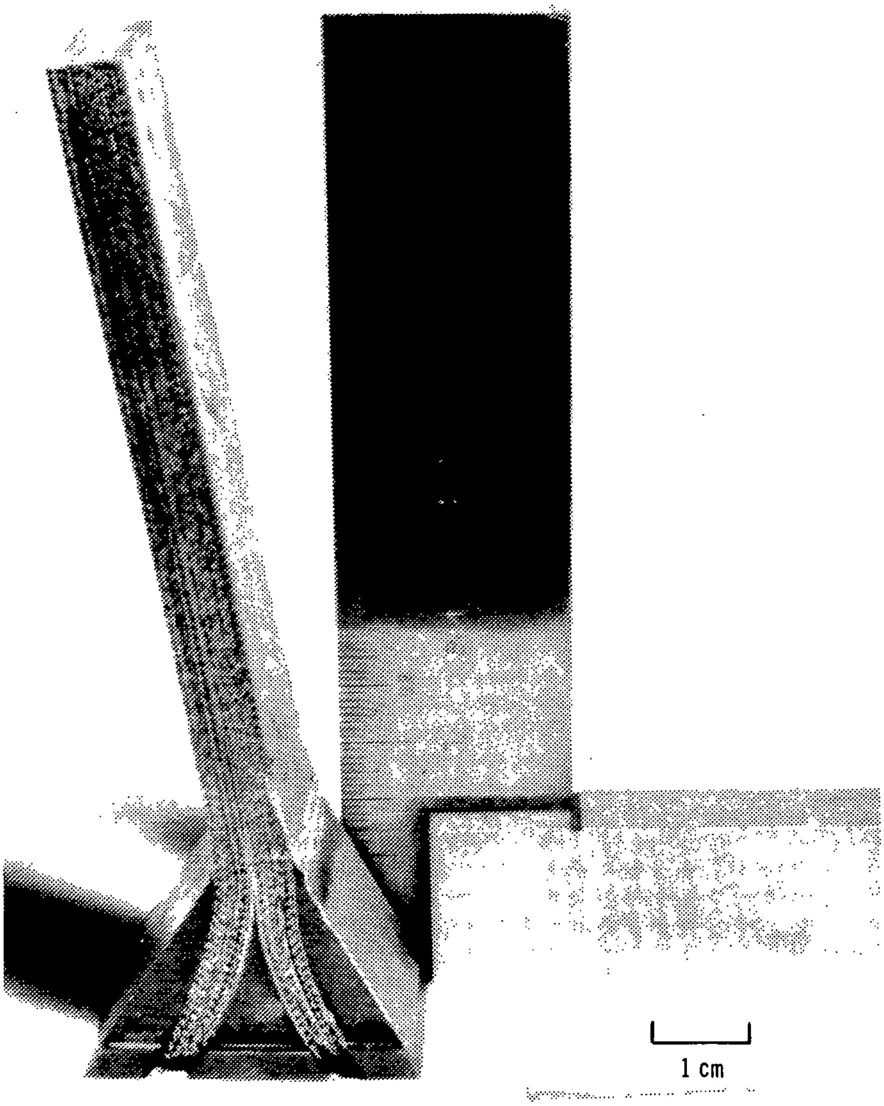


Figure 15. - Boron/aluminum blade-like specimen after high-velocity ballistic impact testing. (From ref. 2.)

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