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IMPROVED IMPACT-RESISTANT BORON-ALUMINUM COMPOSITES FOR USE AS TURBINE ENGINE FAN BLADES

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ABSTRACT

Efforts to improve the impact resistance of B/Al are reviewed and analyzed. Thin-sheet Charpy and Izod impact tests and standard full-size Charpy impact tests were conducted on unidirectional and angleply composites containing 4, 5.6 and 8 mil boron in 1100, 2024, 5052 and 6061 Al matrices. Impact failure modes of B/Al are proposed in an attempt to describe the mechanisms involved and to provide insight for maximizing impact resistance.

The impact strength of B/Al was significantly increased by proper selection of materials and processing. The use of more ductile matrices (1100 Al) and larger diameter (8 mil) boron fibers gave the highest impact strengths by allowing matrix shear deformation and multiple fiber breakage.

Pendulum impact test results of improved B/Al were higher than 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 blades in aircraft gas turbine engines.

INTRODUCTION

Studies by NASA and the Air Force have shown the advantages of using composites as rotating fan and compressor blades in turbine engines. Composites offer lighter weight, lower cost, and higher specific strength and stiffness, resulting in improved engine performance and lower direct operating costs.

Most prior materials development has been directed toward using high specific strength and stiffness composites for airframe structures. High mechanical properties are most important for these applications and little attention has been given to impact resistance.

However for rotating fan and compressor blades in aircraft engine applications, impact and foreign object damage (FOD) resistance become as important to operational performance as strength and stiffness. Ref. 1 defined a foreign object debris spectrum as small-body and large-body damage. Small-body damage includes hard objects such as sand, rocks, rivets, and ice balls. Large-body FOD is caused by hard bodies such as ice slabs, and soft bodies such as birds. Localized damage from small-body impact can result in minor reductions in fatigue strength, while large-body impact may cause complete airfoil separation requiring a reduction in engine speed or complete shutdown.

Collisions with birds are a major flight safety hazard encountered in aircraft operation. Most collisions occur with birds ranging in weight from 4 ounce starlings to 4 pound ducks. During the 1967-69 period, 35% of all aircraft accidents were attributable to bird strikes (ref. 2). About 52% of the bird population (fig. 1) occurs at altitudes less than 500 feet, endangering take-off and landing operations. Take-off conditions are the most severe since the engine is required to operate at full power and power reduction or loss could be catastrophic.

Lack of FOD resistance has been a major obstacle to the use of composites as fan blades in aircraft engines. Composite blades have shown considerable promise in preliminary testing, but in full-stage engine tests, the results have been less than satisfactory. The results indicate that composite blades must have additional impact resistance to become competitive with conventional titanium and stainless steel blades. In addition, root attachment methods used for the blades have caused fiber breakage during fabrication, resulting in premature failure during engine testing.

To overcome these problems, NASA-Lewis has conducted in-house and contractual work to improve impact resistance of both polymer and metal matrix composites for fan blade applications. The objective of this report is to review the programs supporting the impact improvement of B/Al composites and to analyze some of the factors that can increase the impact resistance of metal matrix composites. The results and analysis of the NASA-LeRC in-house programs are presented in greater detail in refs. 3-4, and the contract results in ref. 5. Tensile tests and impact tests on thin-sheet and full-size specimens were conducted to determine the effect of processing variables, matrices, fiber diameters, and angleplies on the impact resistance of B/Al composites. Impact failure modes are proposed and are related to the results obtained.

MATERIALS AND PROCEDURE

Materials Selection

Commercially produced boron fiber, of 0.10 mm (4 mil), 0.14 mm (5.6 mil), and 0.20 mm (8 mil) diameter, was used for composites in this investigation. Because of the standard nomenclature used in the aerospace industry, boron fiber diameter will be referred to in mils, rather than in SI units, throughout this report.

Aluminum alloy matrices, 1100, 2024, 5052, and 6061, were selected to cover a range of impact strengths and ductilities.

Specimen Preparation

All B/Al panels for the in-house study nominally contained 48 volume percent boron and were made by press diffusion bonding of fiber layups between matrix foils. The first series of panels, consisting of 8-ply unidirectional 8 mil B/1100 Al composites, were used to determine the effect of fabrication temperature on impact properties. These panels were bonded at temperatures from 714 K (825 F) to 783 K (950 F).

After selection of a standard fabrication condition of 755 K (900 F) for 0.5 hour at 34 MPa (5 ksi), another series of 1100 Al matrix panels was fabricated. In addition 2024 Al panels were fabricated at 774 K (935 F) and panels with 6061 Al and 5052 Al were fabricated at 805 K (965 F). These panels were also bonded at 34 MPa (5 ksi) for 0.5 hour. Angleply layups were symmetrical from the center. The 8-ply panels were used for tensile and thin-sheet impact tests. Panels for full-size Charpy impact tests were 40-ply for 8 mil boron, 60-ply for 5.6 mil and 80-ply for 4 mil. The full-size Charpy specimens were surface ground to ASTM specifications and a 45-degree notch was cut into one face.

Specimen Geometry

Because of the anistropic properties of composites, specimen geometry must be uniquely defined in terms of fiber direction, pressing direction and notch location. These geometries are shown in fig. 2. The LT, TL, and TT geometries were defined in refs. 6-7. The LT geometry was further defined in ref. 5 as LT, where the testing direction was in a plane normal to the pressing direction, and LT(s), where the notch was on a side parallel to the pressing direction. Tests were conducted on specimens with LT, LT(s), and TT geometries for the studies reported in this paper.

Impact Tests

Three types of pendulum impact tests were conducted: unnotched thin-sheet Izod, unnotched thin-sheet Charpy, and notched full-size Charpy. Thin-sheet tests were conducted because they are more economical in terms of material and machining costs and serve as a convenient screening tool. The cantilever mounting of the Izod test tends to simulate the behavior of a modern, thin-airfoil fan blade in engine operation. Thin-sheet Charpy tests provided an indication of the unrestrained behavior of the material. Full-size Charpy tests provided a comparison of standard specimens with literature values of other materials.

RESULTS AND DISCUSSION

Tensile Test Results

Longitudinal tensile strength of 8 mil B/1100 Al matrix composites decreased with increasing angleply (fig. 3). Longitudinal stress-strain curves, fig. 4, were plotted until the load started dropping, as indicated by the arrows. Unidirectional specimens showed linear behavior to failure. With increasing angleply, nonlinearity and strain to maximum load increased. At failure, the specimens started to separate along the angleply axes.

Transverse stress-strain curves, fig. 5, show that angleplying increased the strain to failure of the 1100 Al matrix composites and increased the transverse strength slightly.

Impact Test Results

Fig. 6 compares the area-compensated LT impact strength of undirectional 1100 Al composites for three different boron fiber diameters. The area under the notch was used for area compensation of standard full-size Charpy specimens. The thin-sheet specimens were unnotched and the entire cross section was used for area compensation. For each type of test, the area-compensated impact strength increased with increasing fiber diameter. The values for full-size Charpy tests of 8 mil boron specimens are shown as a band because the 8 mil B unidirectional panels used for the in-house tests were inadequately bonded and gave excessively low values. Therefore the lower bound represents extrapolations from in-house angleply test results. The upper bound represents impact values from ref. 5. In either case, the increase in impact strength from 5.6 to 8 mil boron specimens is considerably greater with the full-size Charpy tests than with the thin-sheet tests.

The area-compensated full-size Charpy impact strength was much higher than that of thin-sheet specimens. Properly bonded full-size Charpy and unidirectional and low angleply thin-sheet specimens failed by fracture of all fibers in the cross section, with matrix plastic shear prior to fiber failure. Full-size Charpy specimens had more shear than thin-sheet specimens. Higher angleply specimens underwent bending distortion but were pushed through the grips at low impact energies with minimum fiber breakage.

The difference in area-compensated impact strength values for thin-sheet and full-size impact tests is related to their thickness and failure mechanism. Refs. 8-10 reported a transition in fracture and delamination behavior at a thickness of 0.25 cm (0.1 inch). Below this thickness, plane stress conditions applied and delamination stresses were very high. Fiber/matrix bond failure occurred due to shear stress concentration at the notch tip. Above this thickness, plane strain conditions applied where 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. 8).

These results indicate that thin-sheet impact tests can be used as a screening tool to rank impact behavior of various B/Al composites, but the quantitative results of one type of test cannot be extrapolated to another. It should be noted that the comparison of area-compensated results from full-size and thin-sheet specimens appeared to be valid qualitatively despite the fact that the thin-sheet specimens were below the transition thickness of 0.1 inch while the full-size Charpy specimens were above. This probably influenced the inability to extrapolate quantitative values from one test to another. The rankings were consistent for tests on different matrices, fiber diameters, and angleplies where failure ocurred. In angleplies where the thin-sheet specimens deformed, but did not fail, thin-sheet results could not be used for accurate ranking purposes. The indicated impact strengths were actually a measure of: 1) impact strength, if the material were strong enough or brittle enough not to deform excessively, or 2) bending stress, if the material was pushed through the holders without fiber fracture, or 3) a combination of the two, where the material partially deformed and partially failed.

Factors Influencing Improved Impact Behavior of Boron/Aluminum Composites

One of the problems inherent in the evaluation of composite toughness is that a variety of testing methods have been used. Interpretation of results are different depending upon whether notched tensile tests or bending/impact tests are conducted. The ends are rigidly restrained in tensile tests, while in slow bend or impact tests, both ends may be free (Charpy) or one end may be clamped (Izod). Although strength in bending should be comparable to strength in tension, the strain behavior is different. Therefore, interpretation of results and prediction of behavior should be approached with caution when comparing fracture toughness, work of fracture or impact strength results from different types of tests.

Notched Charpy and Izod impact tests are accepted as convenient methods of determining the susceptibility of a material to brittle fracture at high strain rates. Although data from these tests have been used with some success, the approach has been largely empirical (ref. 11). For homogeneous materials, the effects of notch geometry and elastic and plastic deformation under plane stress and plane strain conditions at both the notch region and throughout the specimen are very complex. The stress state and toughness behavior of composites are even more complex because of the divergent properties of the two constituents.

Measurement of fracture energy. - Ref. 12 states that two different concepts can be used to measure fracture energy. One involves measurement of the total energy introduced into a specimen during fracture, averaged over the entire fracture process. This category includes work of fracture and Charpy/Izod impact testing. The other involves measurement of the initial rate of strain-energy release at failure and includes fracture mechanics analyses pertaining to initiation of fracture. Results on carbon fiber reinforced glass (ref. 12) showed that work of fracture, which included fiber failure and fiber pull-out, was much larger than the energy required to initiate fracture.

An empirical relation to predict impact properties of composites was presented in refs. 6-7. Good agreement was reported in the prediction that impact strength of B/Al may be increased by increasing the tensile strength, volume percentage, and diameter of the fiber and by decreasing the shear strength of the matrix. This relation may be valid for predicting general trends, but is probably not valid for exact calculation. The apparent agreement noted in refs. 6-7 may be coincidental.

Results obtained in the NASA-LeRC programs show that the impact energy of B/Al composites also depends upon other factors, related to fabrication conditions and failure mechanisms. This dependence was predicted in ref. 13 where impact energy density (strain energy divided by volume) was shown to be influenced by a correlation coefficient, which is a complex function of constituent properties based upon fabrication history.

Relation of fracture mode to impact energy absorption. - Refs. 14 and 15 reported that work of fracture of composites is influenced by the strength and fracture behavior of the fiber, the matrix, and the interface between the two. Contributions to energy absorption by each are interrelated and can limit or enhance the contributions of the others.

Table I summarizes the relation of fracture mode to impact energy absorption possible in B/Al. The lowest energy absorption would be from cleavage failures. Although not encountered in this program, cleavage failure could occur in overbonded composites where interfacial reaction has forced the fiber to lose its identity and failure would occur in a manner similar to brittle homogeneous materials. A planar fracture would have slightly higher energy absorption. In planar fractures, energy absorption would be primarily controlled by the fiber fracture energy, with no matrix contribution. Delamination or fiber pull-out failures would have medium impact energy absorption. In delamination, energy is absorbed by surface energy release upon delamination of the B/Al or Al/Al interfaces. With fiber pull-out, energy is absorbed by frictional sliding and plastic shear at the interface. Failure by matrix shear with single fiber failures gives high energy absorption because each component makes a contribution to the energy absorbed by the composite. The fiber contribution comes from fiber fracture energy, while the matrix and the interface contributions are by shear displacement energy. Matrix shear with multiple fiber breakage gives the highest impact energy absorption. In this case the fiber contributes additional energy absorption because of multiple breakage and the matrix contribution is increased because of the additional plastic shear allowed.

While the table indicates the relation of failure mode to impact energy absorption, it does not indicate how the toughness of composites can be improved. In this paper, the materials and processing variables that can increase composite toughness by exploitation of these fracture modes will be discussed.

<u>Effect of fabrication temperature</u>. - Impact resistance of B/Al can be increased by the use of fabrication temperatures that allow adequate bonding (to prevent delamination and to make failure dependent upon fiber fracture energy) to obtain properties required for a given application. At the same time, the temperature must be low enough to prevent excessive aluminum boride formation (so that the fibers can exhibit maximum strain to failure).

Area-compensated Izod impact strength is plotted in fig. 7 for thin-sheet specimens bonded for 0.5 hour at various temperatures. Two curves are plotted on this figure: one for delamination failures and the other for fibrous failures. For delamination failures, the impact strength increased with increasing temperature, due to improved bonding with temperature. Fibrous failures did not occur at lower bonding temperatures. Where fibrous failures occurred, the impact strength decreased with increasing temperature.

Specimens fabricated at lower temperatures failed by delamination at low area-compensated Izod impact energies. Bonding was not adequate at these temperatures to allow the composites to attain their full impact strength. The fiber/matrix interface was weak and some specimens even delaminated upon machining prior to testing.

At higher bonding temperatures, the area-compensated thin-sheet Izod impact strength increased. With adequate bonding, the stress to cause delamination at the fiber/matrix interface increased and the matrix could undergo sufficient shear deformation to fracture the fibers. Thus for maximum impact resistance, the failure mechanism changed from being interface controlled delamination to being fiber fracture controlled.

The maximum area-compensated impact strength for B/1100 Al was in the 741-755 K (875-900 F) range in the NASA-LeRC in-house program. Ref. 5 reported that maximum impact properties were obtained using their fabrication cycle at 727 K (850 F). Thus there is probably a range over which maximum impact resistance can be obtained. This range would be dependent upon the complete fabrication cycle used, and upon the foil surface condition and amount of deformation present.

After fabrication at temperatures in excess of 783 K (950 F), the impact strength would probably drop further, due to property degradation from fiber/matrix interfacial reaction. The formation of a thin brittle phase layer at the interface reduced the strain capability of the fiber, thus reducing tensile and impact strength. Although impact data were not obtained from specimens bonded above 783 K (950 F), degradation has been reported by others after processing at higher temperatures. The fatigue limit of B/6061 Al composites was reduced by increasing the bonding temperature (ref. 16). Ref. 17 reported a 20% increase in full-size Charpy impact strength of Borsic/6061 Al composites to 9.4 joules (7.0 ft-lbs) by reducing the bonding temperature from 838 K (1050 F) to 723 K (842 F).

Effect of matrix. - The purpose of the matrix is to provide sufficient ductility to permit the fibers to attain their full strength during the impact process. With sufficient matrix ductility, the fibers can more nearly approach their full strain capability, and failure can occur in an optimum manner where the matrix and the fiber make a full contribution to the fracture energy.

In this program, 8 mil diameter boron fibers were used to reinforce four aluminum alloy matrices: 1100, 5052, 6061, and 2024. These alloys represented different combinations of strength and ductility. Literature data cited for these alloys serve only as an indication of anticipated behavior in composites since the stress-strain behavior is changed by restraint by the fibers. Shear strength becomes an important criterion only if the shear strength of the matrix is lower than the shear strength of the fiber/matrix interface. This was demonstrated in ref. 18, which showed that the fracture toughness of Borsic/1100 Al was independent of B/Al interfacial bond strength.

Ref. 19 proposed that for matrices where the failure strain is higher than that of the fibers, a crack will propagate by sequential failure of the fibers, followed by failure of the matrix along a line joining adjacent fiber breaks. If there is a flaw-dependent length-strength (ref. 20) effect, where the fibers break at different stresses, fiber fractures will not be alined and matrix shearing will occur between fiber failures. This situation is shown schematically in fig. 8-a. Analytical prediction of work of fracture for this case is difficult because of problems in determining the total area undergoing shear. If the strengths of the fibers are uniform and they do not have flaws distributed along their length, the fracture will be nearly planar and the crack will not be deflected from a path directly across the specimen. This would be the case for a plastically deforming fiber such as ductile tungsten wire. Under these conditions, no fiber pull-out would occur and work of fracture would be determined by contributions from plastic deformation of the components. In the case of brittle fibers, such as carbon or boron, fracture is initiated by sequential failure of the brittle fibers on a plane normal to the tensile axis. Ref. 19 states that fracture of brittle fibers should absorb little energy and that the plastic deformation of matrix bridges connecting fiber lengths on either side of the incipient fracture will determine the work of fracture.

For matrices where the failure strain is lower than that of the fibers, failure will be initiated by the growth of a crack in the matrix (ref. 19). This crack will tend to be planar, and unbroken fibers will be left bridging the crack. These fibers will fail eventually at weak points adjacent to the plane of the matrix crack. The matrix fracture surface will be smooth with some surface depressions and projecting pulled-out fibers. This situation is shown in fig. 8-b. In this case, work of fracture can be predicted using the analysis of ref. 20.

Refs. 14 and 19-22 reported that maximum work of fracture occurs with discontinuous fiber composites. When a crack passes through a composite, fibers shorter than the critical length are pulled out from the matrix, rather than broken. Fibers of the critical length have a maximum distance of pull-out. Fibers longer than the critical length will fail in tension, normally at a lower work of fracture. Work of fracture is thus a combination of the work needed to debond the fibers from the matrix and the work done in pulling the fibers out of the matrix. However, it should be emphasized that this occurs primarily in the case where the matrix is more brittle than the fibers (ref. 19).

For the case where the fiber is ductile and the matrix is very brittle, fracture would be initiated in the brittle matrix. Multiple cracking of the matrix would occur because deformation is not limited to the plane of final fracture.

Results of this program follow the behavior outlined above. Thin-sheet Izod and Charpy, as well as full-size Charpy impact strength of B/Al was increased using more ductile and weaker matrices. Composites with 1100 Al matrices had significantly higher impact strengths than those with other matrices. Composites with stronger and less ductile matrices had the lowest impact strengths. Similar results were reported in ref. 5. The fracture surface became more jagged and irregular with increasing impact strength, and fiber/matrix projection zones of fibers connected by bonded matrix were projecting out of the fracture surface. Fig. 9 shows comparisons of fracture surfaces for B/Al composites observed in ref. 5. For 5.6 mil B/1100 Al composites (fig. 9-a), some bare fiber pull-out can be seen at the tops of some of the projection zones, but the general jaggedness and projection zone formation is apparent. Fig. 9-b shows that the projection zone effect is more pronounced with the higher impact strength 8 mil boron composites. Fig. 9-c shows the brittle, planar fracture surface of a lower impact strength 5052 Al matrix composite with no fiber/matrix projection zones present.

Fig. 10 shows failed full-size Charpy specimens. The low-energy fracture of the 5.6 mil B/5052 Al composite (fig. 10-a) was planar with no matrix shear. Restraint by the boron fibers reduced matrix ductility below its unreinforced value. The 2024, 5052, and 6061 Al matrix composites acted in the matrix-less-ductile-than-fibers case of ref. 19. The ductility of the 1100 Al matrix was sufficiently high to be more ductile than the fibers. The higher-energy 5.6 mil B/1100 Al composite (fig. 10-b) shows a jagged fracture surface with a large amount of shear deformation. Fig. 11 shows that the shear displacement at the ends of failed LT full-size Charpy specimens from ref. 5 increases linearly with increasing impact strength. This deformation increased the impact strength of the composite in two ways. First, additional energy was absorbed through multiple breakage of the fibers. Second, the matrix absorbed more energy through additional shear after the initial fiber failures.

In high impact strength B/1100 Al composites, the matrix sheared during pendulum impact and the fibers failed in tension. With the additional matrix shear allowed by the ductile 1100 Al matrix, the tensile stresses in the intact portions of the broken fibers continued to increase and failed the fibers again. Composites with 8 mil boron showed more matrix shear ductility and multiple fiber breakage. Fig. 12 shows a failed 8 mil B/1100 Al thin-sheet Izod specimen. The outer fibers have radial cracks in the fracture region at fairly regular distances along the fiber length. This indicates that multiple fiber breakage occurred prior to and during failure. This multiple fiber breakage was localized in the fracture region.

<u>Effect of fiber diameter</u>. - Area-compensated LT impact strengths of 1100 Al matrix composites with various fiber diameters are shown in fig. 6 for three types of impact tests. These results indicate that the impact strength of B/Al increased with increasing fiber diameter. Ref. 5 also reported that the impact strength of B/1100 Al was higher using 8 mil boron than with 5.6 mil. Limited data in refs. 6 and 17 showed similar trends. Work of fracture for copper matrix composites with brittle, recrystallized tungsten wires also increased with increasing fiber diameter (ref. 10).

For a given fiber content, increasing fiber diameter decreases the total surface-to-volume ratio of the fibers within the composite. Increasing the diameter from 4 to 5.6 mils, or from 5.6 to 8 mils doubles the cross-sectional area of a single fiber, but only increases the shear area by 40%. The shear stress would be higher at a given tensile load, allowing a ductile matrix and/or fiber/matrix interface to yield and shear prior to composite fracture. Shear is desirable if the matrix has sufficient ductility to allow plastic shear without premature crack initiation prior to fracture.

Interfiber distance must be great enough to allow the matrix to exhibit its full ductility and to absorb impact energy by shear deformation. The increase in effective fiber diameter caused by restraint of the matrix by the fibers (ref. 23) reduces the distance between adjacent fibers for accommodating shear displacement. This effect decreases with increasing fiber diameter since interfiber distances are correspondingly larger for a given fiber content. Specimens with 4 mil boron displayed little shear during fracture and had the lowest impact strengths. No multiple fractures were observed and the ductility of the 1100 Al matrix was minimal. The increase in effective fiber diameter reduced the already small interfiber distance even further and the matrix could not act in a ductile manner.

Increasing the boron diameter to 5.6 mils increased the interfiber spacing. These specimens exhibited increased fracture ductility and impact strength. In this case the interfiber spacing was sufficient to allow some shear and multiple breakage.

Comparison of figs. 9-a and 9-b shows that the 8 mil boron specimens had much more pronounced fiber/matrix projection zones. This can be attributed to the interfiber distances being large enough to allow the matrix to achieve sufficient ductility to maximize fracture energy, through additional shear and subsequent multiple fiber fractures. Comparison of figs. 10-b and -c shows the increases in shear deformation during impact of 1100 Al matrix composites allowed by increasing the boron fiber diameter from 5.6 to 8 mils. The use of 8 mil boron in composites with other matrices also increased their impact strengths over those previously reported for 4 mil boron. From these results, it may be postulated that the use of even larger diameter boron fibers could further increase the impact strength of composites with 6061 and 5052 Al matrices.

Ref. 24 reported results from Charpy impact tests on boron, carbon, or glass fiber composites with resin matrices of various toughness. Calculations were made to determine the relative contribution of fiber pull-out, shear delamination, and fiber fracture energies. Two-thirds of the calculated energy came from the energy absorbed by fiber fracture, which was in turn proportional to the area under the stress-strain curve of the fiber. Glass fibers, with much higher strength and failure strain, had the largest area under the stress-strain curve and gave the highest Charpy impact strengths. Boron fibers were next, and carbon fibers, with the lowest strain and area under the curve had the lowest impact results. Furthermore impact strength was independent of the toughness properties of the matrices due to the overpowering influence of the fibers.

These results are significant because they show that in a brittle matrix system, the major contribution to energy absorption comes from fiber fracturing. Composite impact properties are an interaction of the energy contributions of each constituent: the matrix, the fiber, and the interface. However the strain and impact behavior of each component are interrelated and must be such that the full contribution from each can be attained. Brittle resin matrices do not contribute much to the energy absorbing capability of a composite. A ductile matrix, such as 1100 Al, can make a significant contribution to the overall impact energy by absorbing additional energy by matrix shear as well as by allowing multiple fiber fracture. Thus it is vitally important to have a matrix with sufficient ductility to allow the fibers to attain a greater portion of their full strength and strain capability.

Effect of angleply. - Due to the anisotropic nature of composites, the transverse properties of unidirectional composites may not be high enough to withstand stresses encountered during component service. Angleply layups can be used to improve the transverse properties; this transverse improvement, however, is attained with a considerable penalty in longitudinal properties.

Angleplies of ±7, ±15, ±22, and ±30 degrees for 8 mil B/1100 Al composites were studied. In addition, results from tensile and full-size Charpy impact tests for three angleplies with 1100 and 5052 Al matrix composites were reported in ref. 5. The first angleply was (±45/0), consisting of 50% unidirectional plies in the central core with 25% alternating ±45 degree shells on each side of the core. The second was (0/±22)nT, and consisted of repetitive 0,+22,-22 degree plies. The third was alternating ±15 degree plies.

The longitudinal tensile strength of B/1100 Al decreased with increasing angleply. This reduction was caused by a decrease in the elastic strain range and an increase in nonlinear strain shown in the stress-strain curves of fig. 4. The transverse modulus and strength increased slightly with increasing angleply. Similar results for B/1100 Al and B/5052 Al composites were reported in ref. 5, but with higher longitudinal and transverse strengths due to different bonding conditions.

Fig. 13 compares the area-compensated longitudinal impact strengths of angleplied B/Al composites from the NASA-LeRC in-house program and from ref. 5. Full-size Charpy specimens showed a linear loss in impact strength with angleply. Because of the difference in bonding conditions, the two sets of data are displaced from, but parallel to, each other. Thus, the trends from the two can be compared. Unidirectional specimens had higher impact strengths than any of the angleplies. The reduction was minor up to ±15 degrees. The ±7 degree angleply had a minor loss in impact and tensile strength compared with the unidirectional specimens. Increasing angleply decreased the impact strengths where fracture occurred. At angleplies greater than ±15 degrees, the non-linear stress-strain behavior and low strength allowed the composites to deform without applying sufficient stress on the fibers to attain high impact. The angleply specimens that did not break, ±22 and ±30 degree, underwent considerable stretching and distortion during impact testing and showed a large amount of shear, but the stresses required for deformation were low due to the low flow stress of these composites. The fibers were not strained enough to make their maximum contribution to the properties of the composite. The maximum angleply that allowed the fiber properties to be utilized was ±15 degrees. In this case, the fibers fractured after attaining sufficient strain to give high stresses and impact energies.

The ±45 shell-0 degree core configuration had the best transverse strength and impact properties, but also had the lowest LT properties (ref. 5). The ±22, 0 angleply gave slightly lower TT impact and tensile strengths than the ±15 degree angleply. This was due to the 0 degree fibers which gave adverse results in the transverse direction. The best combination of longitidinal and transverse impact and tensile results was obtained with the ±15 degree angleply.

Effect of matrix enhancement. - Another method of improving transverse strength 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.

Refs. 6-7 reported that the addition of 6 v/o stainless steel wire, oriented in the transverse direction, increased TT Charpy impact strength of 4.2 mil Borsic/6061 Al composites from 1.5 to 6.0 joules (1.1 to 4.5 ft-lbs). Further work (ref. 13) showed that LT Charpy impact strength of 4 mil B/6061 Al was increased by 60% to 26 joules (19 ft-lbs), with an accompanying increase in transverse tensile strength, by using a dual alloy matrix of 6061/1100 Al. It was suggested that LT and TT impact strength and transverse tensile strength could be increased further by using titanium foil as matrix enhancement.

Results were reported in ref. 25 for diffusion bonded and adhesively bonded 5.6 mil B/6061 Al composites, as well as 5.6 mil B/6061 Al hybrid composites with adhesively bonded 0.038 mm (0.0015 in.) thick Ti-6Al-4V foils. The area-compensated thin-sheet Izod impact strength of adhesively bonded B/Al was increased from 32 to 43 joules/sq m (15 to 21 ft-lbs/sq in.) for the B/Al+Ti hybrid. Without hybridization, values for diffusion bonded 4 mil B/6061 Al were 45 joules/sq m (22 ft-lbs/sq in.) and 49 joules/sq m (24 ft-lbs/sq in.) for diffusion bonded 5.6 mil B/6061 Al. Further hybridization by adding graphite fiber/epoxy plies to B/Al+Ti hybrids increased area-compensated Izod values to 117 joules/sq m (56 ft-lbs/sq in.). (These values should be compared to the thin-sheet Izod results reported in this paper: 8 mil B/1100 Al: 192 joules/sq m; 5.6 mil B/1100 Al: 89 joules/sq m; and 4 mil B/1100 Al: 75 joules/sq m).

Ref. 5 reported the use of Ti-6Al-4V foils with 5.6 and 8 mil B/1100 Al to determine the effect of matrix enhancement. Results showed that matrix enhancement reduced longitudinal tensile strength 15% and reduced full-size LT Charpy impact strength by over 50% for both composites. The transverse tensile strength was increased from 65 MPa (10 ksi) to 266 MPa (39 ksi), however TT Charpy impact strength was only increased from 1.4 joules (1 ft-lb) to 4.1 joules (3 ft-lbs). This slight increase in TT impact strength did not justify the sacrifice in LT impact. The same trends held for angleply B/5052 Al composites. The LT impact strength was reduced by more than 50% while the TT impact strength was virtually unchanged by titanium foil enhancement.

The data obtained in ref. 5 seem to differ from other reported results. It is generally thought that titanium foil interleaves should increase impact strength of B/Al. Titanium is very impact resistant in the unnotched condition, however a notch reduces the full-size Charpy impact strength from 318 joules (220 ft-lbs) to 23 joules (15 ft-lbs). Thus in a notched impact test, titanium foil matrix enhancement should only improve impact strength of composites having impact strengths below that of notched titanium (23 joules). Titanium foil restrains the matrix from shearing, thus making fracture and crack initiation more difficult, thereby increasing impact strength of brittle composites. It also provides delamination planes for low impact composites, which rely on delamination surface energy dissipation to improve impact behavior. The matrix ductility restraint imposed by matrix enhancement foils, however, 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.

Comparison of the SEM fractograph presented in fig. 14 with that in fig. 9-b shows that the fiber/matrix projection zones are broken up by the titanium foils. The fracture is planar with much bare-fiber pull-out and no evidence of matrix shear ductility.

Effect of directionality. - An unexpected directionality effect reported in ref. 5 was the reduced impact strength observed in full-size Charpy tests in the LT(s) direction. The impact strength for the LT(s) geometry dropped as much as 30-50% below the LT strength.

In diffusion bonding, matrix foils are placed between fiber layers and consolidated. Upon impact testing of LT specimens, the crack must propagate sequentially through fully dense aluminum foils with weaker Al/Al interfaces separating the individual foils. In the other case, LT(s), the crack must propagate simultaneously across the entire number of plies acting as a unit.

If bonding were not perfect, the strength of the foils, in the fully dense direction in the plane of the foil, would probably be greater than the strength in the direction where the foils were bonded to each other. This can be seen from the notch-initiated delamination present in full-size LT Charpy specimens (fig. 15-a). After delamination, the specimen bends by shear and acts in a ductile manner, resulting in high impact energies. The opposite case, LT(s), does not undergo this type of delamination below the notch (fig. 15-b). SEM fractographs of high-energy LT specimens, fig. 9-b, showed massive fiber/matrix projection zones. The LT(s) specimens (fig. 16) showed less fiber/matrix projection zone formation. The fibers are alined in intact vertical planes and appear to show evidence of bare-fiber pull-out. The vertical planes are from the individual ply layup during consolidation. The crack propagation direction is normal to the edge of the ply and the fracture crack proceeds throughout all the plies simultaneously. Instead of having uniform plies to deform sequentially by shear, LT(s) specimens must fracture simultaneously through all the plies. Since none of these plies are oriented preferentially for shear, the matrix cannot shear and the fibers are not permitted to exhibit their maximum strain capability. Thus, the impact strength of LT(s) specimens is reduced to that approaching a restrained, non-ductile matrix.

Application of Improved Impact Technology to Aircraft Gas Turbine Engine Fan Blades

The very large increase in pendulum impact strength of improved B/Al composites described in this paper is very encouraging. The advantages of the improved B/Al composites are shown in fig. 17, which compares current values with impact strengths of previous B/Al and notched titanium. These results provide a basis for expecting that a significant improvement in fan blade performance might be obtained. 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.

Blade-like shapes were fabricated, tested and reported in ref. 5. These blade-like specimens had a flat, untwisted airfoil-like section and a splayed 3-wedge root. The root was placed in a clamp and the specimens were subjected to low-velocity impact tests. Specimens of ±15 degree angleply 8 mil B/1100 Al failed at the root-airfoil fillet after considerable shear, thus indicating that the matrix shear displacement took place in a manner similar to that observed in Charpy/

Izod thin-sheet and standard specimens. A limited number of high-velocity tests with blade-like specimens were also performed. Fig. 18 shows a B/Al specimen after high-velocity ballistic impact with RTV silicone rubber simulated birds (ref. 5). Specimens were able to withstand impact energies up to 250 joules (184 ft-lbs), the maximum energy tested. 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.

Both low-velocity pendulum and high-velocity ballistic impact results are encouraging. Additional tests 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 broad application to fan blades.

A start has been made with this effort and the results obtained thus far are very encouraging. These promising results should serve to further the continuation of the development of B/Al composites to obtain the large pay-off in performance gain, fuel economy, and cost and weight reduction that composite materials can provide when applied to fan blades for aircraft gas turbine engines.

SUMMARY OF RESULTS AND CONCLUSIONS

The following results and conclusions were obtained from studies to improve the impact properties of diffusion bonded B/Al composites:

1. Pendulum impact test results of improved B/Al were higher than notched titanium and appear to be high enough to give sufficient foreign object damage protection for consideration of B/Al for application to fan and compressor blades in aircraft turbojet engines.

2. Impact strength of B/Al can be improved by proper choice of fabrication temperatures. Processing at below optimum temperatures causes impact strength to be reduced by B/Al or Al/Al interface delamination. Above the optimum, impact strength would be reduced by excessive reaction at the fiber/matrix interface. In this case the bond strengths are in excess of those required for best impact resistance.

3. Impact strengths of composites with an 1100 Al matrix are significantly higher than with 2024, 5052 and 6061 Al matrices. More ductile matrices allow additional energy absorption through shear deformation and multiple fiber breakage.

4. Larger diameter boron fibers increased impact strength. They provide larger interfiber spacing, allowing the matrix to act in a more ductile manner and permit the fibers to attain a greater portion of their full strength and strain capability.

5. The LT(s) impact strength (notched side parallel to pressing direction) was lower than the LT impact strength (notched side normal to pressing direction).

6. Transverse tensile and impact properties can be increased through the use of angleply fibers. The optimum angleply for impact resistance appeared to be about ± 15 degrees.

7. Matrix enhancement, using titanium foil interleaves, reduces the longitudinal impact strength of ductile, high impact strength B/Al composites.

8. Thin-sheet Izod and Charpy impact tests can be used for ranking purposes to compare impact properties with full-size Charpy tests, but the quantitative results of one type of test cannot be extrapolated to another.

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TABLE I. - IMPACT FRACTURE MODES IN B/Al COMPOSITES

FRACTURE MODE	ENERGY ABSORPTION			
	TOTAL	CONTRIBUTION		
		FIBER	MATRIX	INTERFACE
CLEAVAGE	V. LOW	V. LOW	V. LOW	V. LOW
PLANAR	LOW	LOW	LOW	LOW
DELAMINATION	MEDIUM	LOW	LOW	MEDIUM
PULL-OUT	MEDIUM	LOW	MEDIUM	LOW
MATRIX SHEAR				
SINGLE FIBER FAILURE	HIGH	HIGH	HIGH	HIGH
MULTIPLE FIBER FAILURE	V. HIGH	V. HIGH	V. HIGH	V. HIGH

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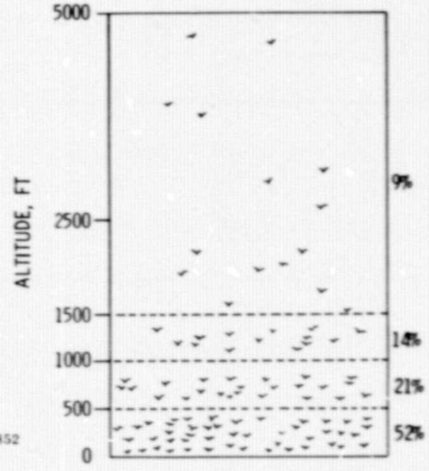

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Figure 1. - Average density of local bird population with altitude (ref. 2).

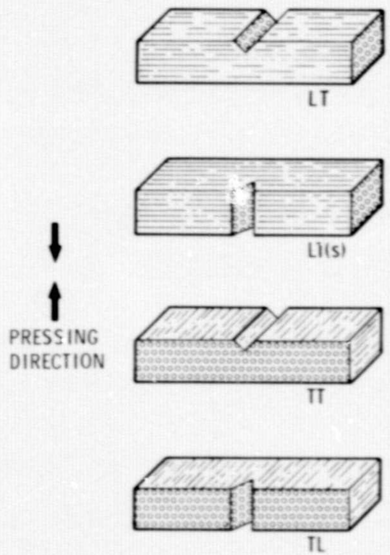

Figure 2. - Charpy impact test specimen geometries.

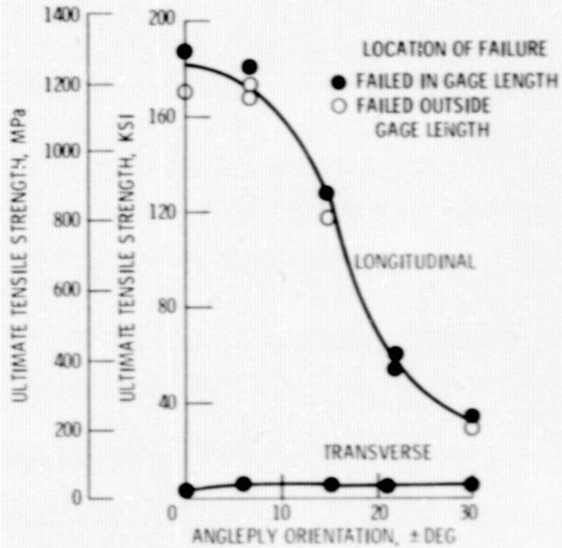

Figure 3. - Tensile strength of 8 mil B/1100 Al angleply composites.

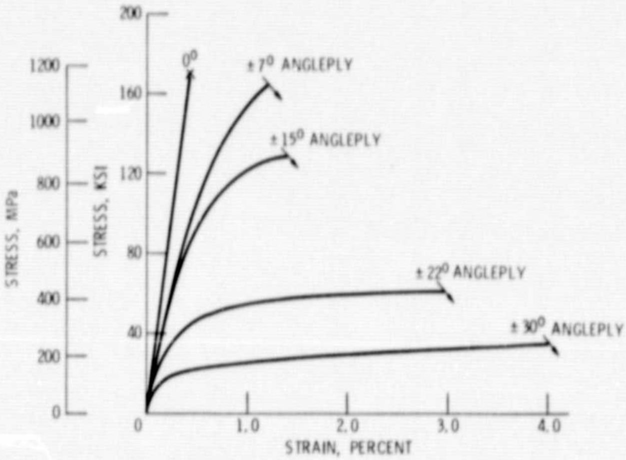

Figure 4. - Stress-strain curves of 8 mil B/1100 Al composites, tested in the longitudinal direction.

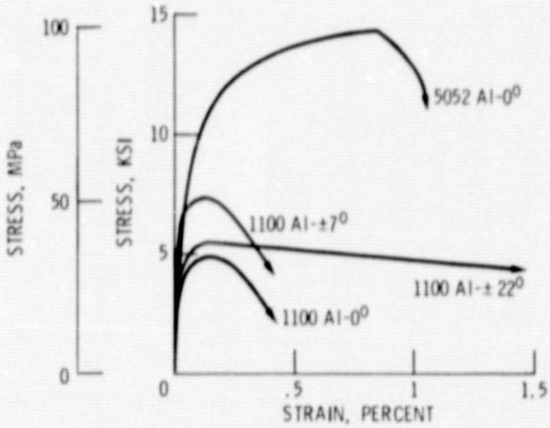

Figure 5. - Stress-strain curves of 8 mil B/Al composites, tested in the transverse direction.

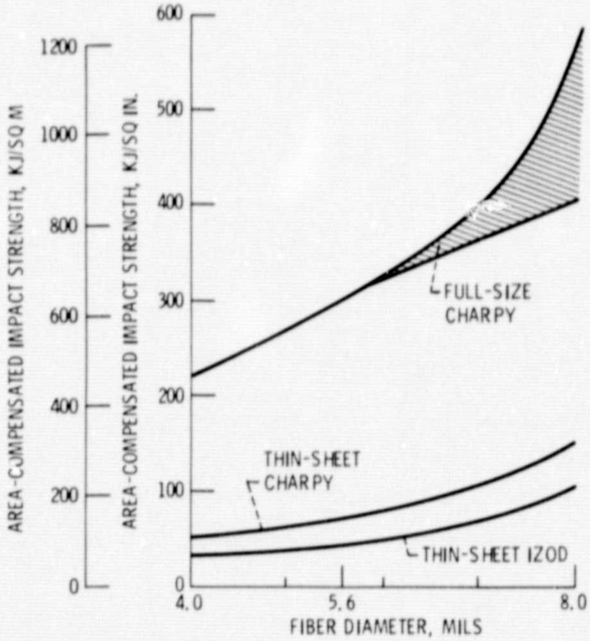

Figure 6. - Area-compensated longitudinal impact strengths of unidirectional 1100 Al composites from different types of impact tests.

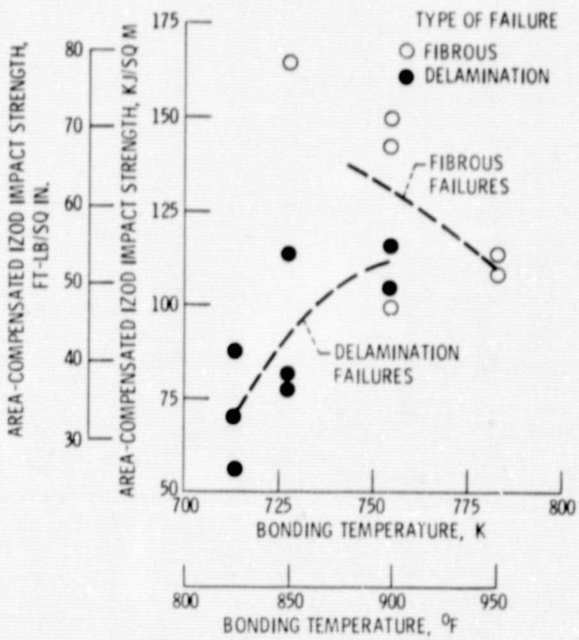

Figure 7. - Effect of bonding temperature on area-compensated thin-sheet longitudinal izod impact strength of 8 mil B/1100 Al composites.

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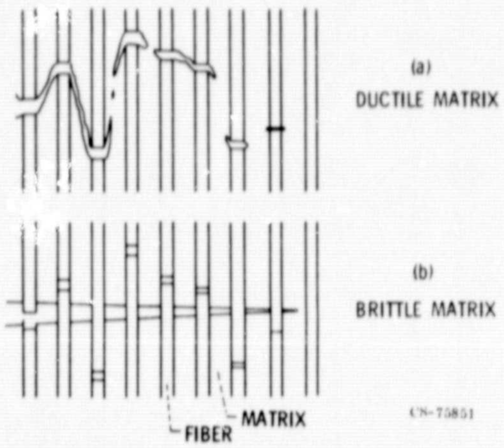
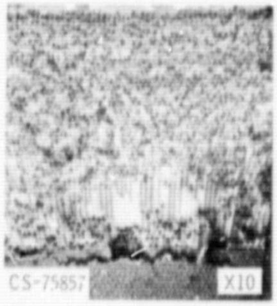
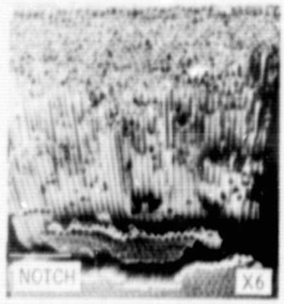


Figure 8. - Crack propagation in composites with matrices of different ductilities (ref. 19).

(a) 5.6 MIL B/5052 Al.

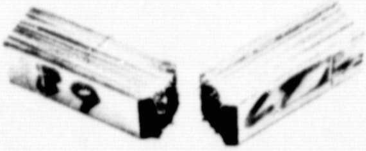


(b) 5.6 MIL B/1100 Al.

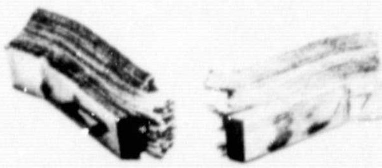


(c) 8 MIL B/1100 Al.

Figure 9. - Failed full-size LT Charpy impact test specimens showing effect of various matrices and fiber diameters on projection zones in B/Al composites. (Ref. 5.)



(a) 5.6 MIL B/5052 Al
18 J (13 FT-LB).



(b) 5.6 MIL B/1100 Al
64 J (47 FT-LB).

(c) 8 MIL B/1100 Al
96 J (71 FT-LB).

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Figure 10. - Failed full-size LT Charpy impact test specimens. (Ref. 5.)

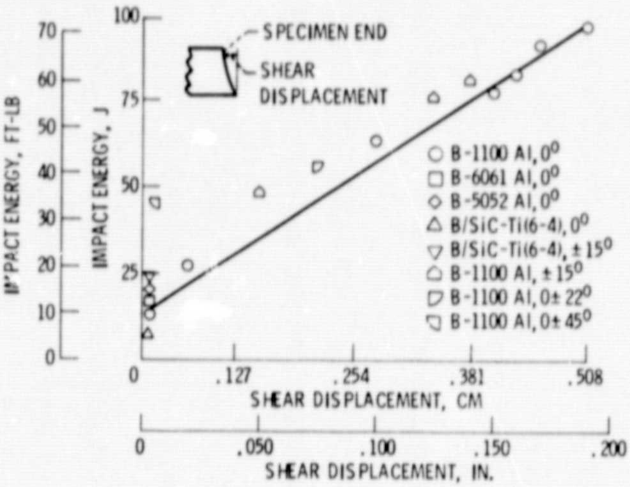

Figure 11. - Shear displacement at the ends of various failed full-size LT Charpy impact test specimens (ref. 5).

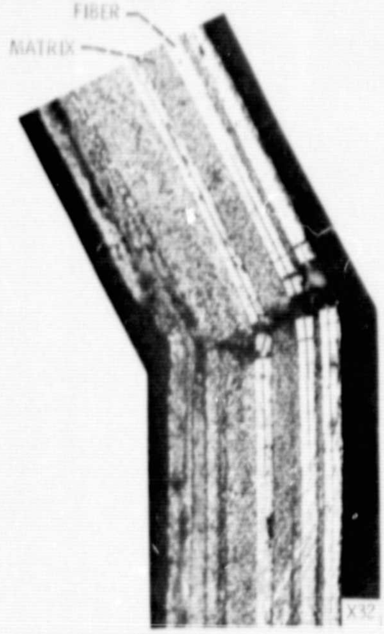

Figure 12. - Fracture region of failed unidirectional 8 mil B/1100 Al thin-sheet Izod impact test specimen showing multiple fiber breakage.

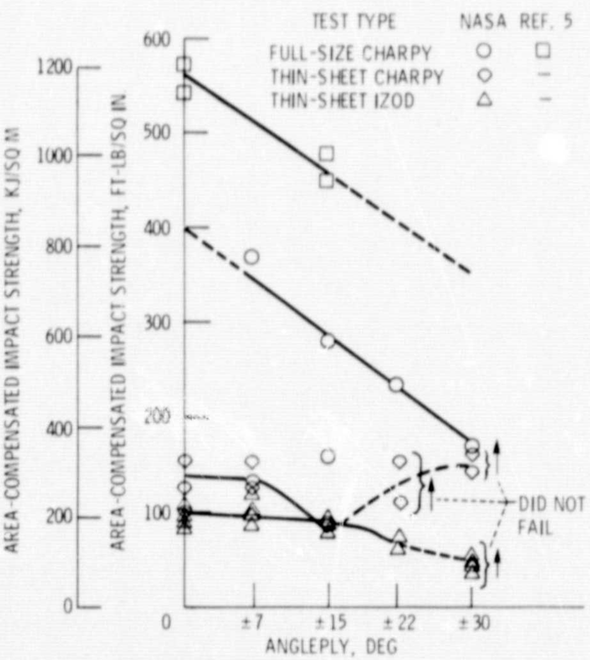

Figure 13. - Area-compensated longitudinal impact strengths of 8 mil B/1100 Al angleply composites.

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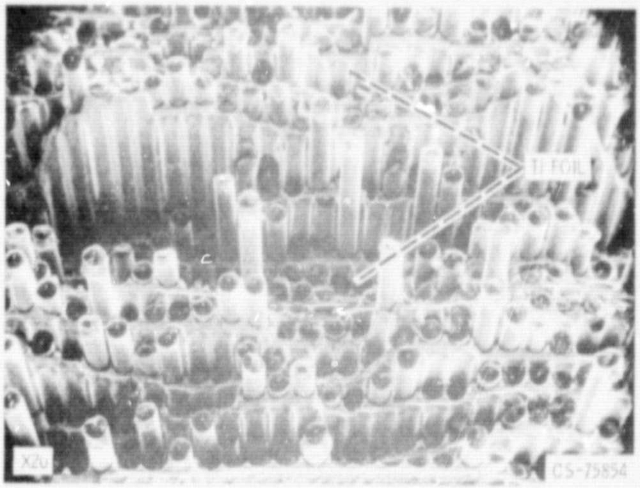
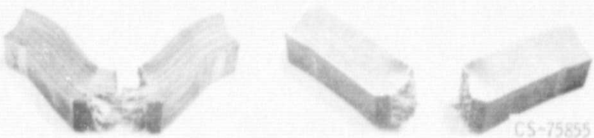


Figure 14. - SEM fractograph of fracture surface of 8 mil B/1100 Al composite with titanium foil matrix enhancement. LT impact strength, 41.5 J (30 ft-lb). (Ref. 5.)



(a) LT; IMPACT STRENGTH, 96.5 J (71 FT-LB).

(b) LT(s); IMPACT STRENGTH, 47.5 J (35 FT-LB).

Figure 15. - Failed full-size LT and LT(s) Charpy impact test specimens. (Ref. 7.)

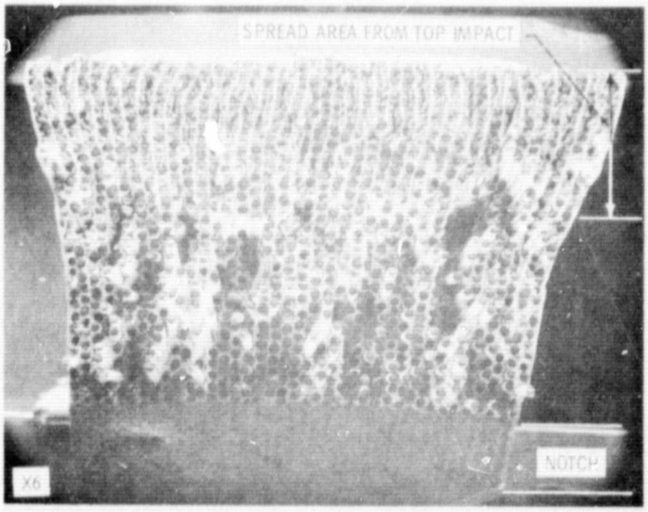


Figure 16. - Fracture surface of failed 8 mil B/1100 Al full-size LT(s) test specimen. Impact strength, 47.5 J (35 ft-lb). (Ref. 5.)

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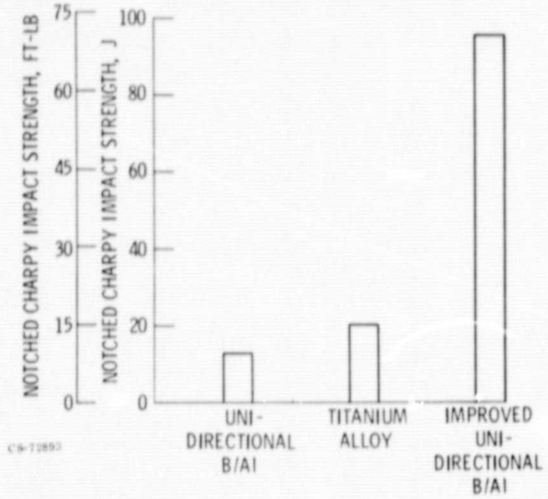

Figure 17. - Progress in impact improvement.

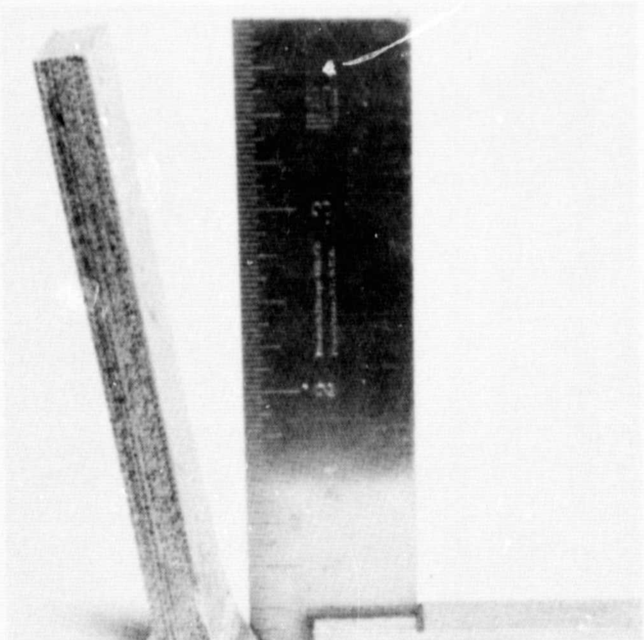

Figure 18. - B/Al blade-like shape specimen after high-velocity ballistic impact testing. (Ref. 5.)