

FAULT TREE ANALYSIS OF FAILURE IN
BORON-ALUMINUM COMPOSITES

by

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I. INTRODUCTION

Fiber reinforced composite materials, consisting of continuous high strength, high modulus fibers in a relatively low strength, low modulus, ductile matrix material, have considerable application as engineering materials. Fibrous composite materials are especially important in engineering structures where a material with high stiffness and high specific strength is needed and where exposure conditions can be severe. As engineering materials become more advanced, engineering applications more demanding, and environmental conditions more aggressive, the need for developing techniques and procedures for realistically assessing the reliability and safety of these materials becomes more pronounced. Unfortunately in some cases assessment comes only after the engineering structure has failed. In any analysis of failure, it is important to determine the events which lead to degradation of the material and determine which causal events precipitated the failure. It is also important to distinguish between causal events and subsequent events, and to determine whether such events are concerned with the properties of the material and/or the nature of the environment. Fault tree methodology, which has been used for systems safety analysis in the aerospace and other

industries is beginning to find some application in the area of failure analysis of engineering structures. Fault tree analysis uses graphical techniques to describe the logical cause and effect relationships between events. Fault tree methodology can be combined with the data obtained in failure analysis to determine the events which lead to the failure of an engineering structure. Failure analysis has been defined as "a technique by which facts are gathered and studied to determine the cause of an equipment or part failure so that preventive action may be taken " (1). One aspect of failure of composite materials analysis is the attempt to relate properties of the components of the composite such as strength and ductility to environmental exposure conditions such as thermal treatment, temperature, applied and residual stress states, fabrication processes, etc., so that the understanding of the failure process is increased. Fault tree analysis aids this attempt by providing a logical system for relating the events which lead to failure in a cause and effect manner.

The high degree of sophistication in modern advanced composite materials can cause unique problems for the analyst involved in the investigation of the failure process. In a simple tension test, a multidirectional fibrous laminated composite can be strained to fracture

with the final fracture involving one of several mechanisms or a combination of several mechanisms. It is not always easy to relate these mechanisms to the more basic events which precipitated the failure. Additionally, much of the physical evidence for the fracture events is frequently destroyed in the failure process or altered by post failure specimen preparation. Fault tree methodology allows the analyst to use inductive reasoning to determine plausible occurrences and to assign probabilities to these occurrences even though physical evidence for their presence may not be obtainable.

The failure process in unidirectional, continuous fiber composite systems has been described by many investigators. Although many of the developments which lead to failure of a composite are described, it is often not clear how the various events which are of importance to the failure process interrelate in a logical sequence. Additionally, some descriptions of composite failure are inadequate because relevant events such as specific material properties or mechanical history are missing from the analysis. Also lacking in many failure descriptions is a technique for relating the events which lead to failure in a logical, cause and effect manner. Fault tree methodology has been applied with success in several failure studies, some of which involve composite materials systems.

II. BACKGROUND DEVELOPMENTS

The failure analysis of a composite system using fault tree techniques requires both a high degree of understanding of the performance and fracture modes of the system being investigated and an understanding of fault tree techniques. A significant portion of the background development has involved the review of fault tree methodology. Additionally, a study of the performance and failure mechanisms of several composite systems has been outlined and several papers which applied fault tree techniques to the failure analysis of engineering systems have been summarized.

A. Use of Fault Tree Techniques

Fault tree methodology is a developing and ongoing science. The extent and scope of its application have not been fully realized. A recent overview of fault tree methodology and its relation to reliability and safety assessment can be found in "Reliability and Fault Tree Analysis"(2). Fault tree analysis dates back to the early 1960's when it was developed as a tool for reliability assessment of the minute-man launch system. Subsequently it has been used for reliability and safety assessment in a number of processes and industries. Barlow and Lambert have described the basic concepts, techniques and appli-

cations of fault tree analysis (3). Their description encompasses the construction of fault trees, the nature, types and uses of events and logic gates, as well as the evaluation and assessment of fault trees, both qualitatively and quantitatively. Powers, Tompkins and Lappe have described the relationship between boolean mathematical models and fault tree construction and have applied their results to safety analysis studies in the chemical process industry(4). In this analysis the events which could cause a fault event or hazard are determined by solving a series of boolean equations which simulate the process. The solution to the series of equations becomes the fault tree.

Fault tree methodology has also been applied to safety and reliability problems in the nuclear industry. Cummings used fault tree techniques in safety evaluation of pressurized water reactors (PWR) to determine the probability of excessive leakage of containment atmosphere following a loss of coolant accident (5). This particular reliability analysis was part of the Rasmussen study on nuclear power plant risks. The scope of the analysis was to gather and organize information and construct a fault tree for a pressurized water reactor.

B. Failure Analysis of Composite Materials

The tensile failure of a unidirectional composite material is a rather complex phenomenon. Failure is very dependent on the individual properties of the components, the types and degree of bonding between the components, and upon the thermal and other environmental conditions to which the composite has been exposed. The strength of many fibrous composite materials cannot be characterized uniquely because the fiber material is brittle (thus fiber strength must be determined statistically), and because fiber-matrix interactions alter the strength characteristics of the composite. Rosen treated fibers in a composite by considering a statistical distribution of flaws and defects along the fiber length. Propagation of fiber defects leads to fiber failures at various stress levels because of the variations in initial defect size (6). Composite fracture occurs when, due to the increasing load, an increasing number of fiber fractures accumulate and a sufficient number of ineffective fiber lengths combine to produce a weak cross section in the composite. Failure of the composite then occurs by tensile fracture of the fibers and matrix. The onset of fiber degradation occurs at strains well below the failure strain of the material. Thus, prior to failure, a number of discontinuous fibers exist in a composite which was initially composed of

continuous filaments. The mechanical properties of the matrix and the nature of the fiber-matrix interactions become increasingly important as the number of fiber breaks increases in the failure process.

Rosen's model for composite failure was improved by Zweben (7) and later by Zweben and Rosen (8) by accounting for the effects of load concentrations in the area of a broken fiber. Rosen assumed the load in a broken fiber to be redistributed over the entire cross section in which the original break occurs. However Zweben assumed that the fibers immediately adjacent to the broken fiber are subjected to load intensities greater than those which are sustained by fibers remote from the fracture site. Thus each fiber fracture site can be considered as a nucleus for the propagation of fiber fracture and the increase in fiber fractures is cumulative in nature. Zweben's improvements of Rosen's model brought predictions of composite performance into closer agreement with experimental results. Zweben also considered the case in which fiber fracture can occur in a non-cumulative manner. In the non-cumulative mode, failure is caused by the breakage of the first fiber which initiates a matrix crack at the fiber break site and the crack propagates through the matrix breaking the fibers in its path and the composite fails catastrophically.

The effect of thermal excursions on the mechanical properties of a composite material is a significant parameter to consider in the failure analysis. Herakovich, Kennedy, and Tenney, in a study to determine the effects of cyclic loading and temper condition on the tensile behavior of boron-6061 aluminum composites, show that thermal treatment to any condition, other than the as fabricated condition, degrades the mechanical properties of 0°, unidirectional boron-aluminum composites (9). The strength reduction is believed to be the result of fiber degradation which results from aluminum-boron reactions at elevated temperatures. This result contrasts conditions in lamina where matrix dominated properties can be assumed. When properties are matrix controlled, thermal excursions have been shown to improve composite properties by strengthening the precipitation hardenable aluminum matrix. However, Grimes, Lad, and Maisel have also studied the effects of thermal cycling on the strength of unidirectional boron-aluminum composites and attribute strength losses to chemical reactions between the fibers and matrix and to the residual stress states which result from the difference in coefficients of thermal expansion in the fiber and the matrix (10). Results indicate that thermal cycling of composite materials tends to have a significantly more damaging effect on the tensile properties

of the unidirectional composite than heating continuously for equivalent times at temperature. Theoretical calculations by Grimes et al indicate that the thermally induced stresses are insufficient to break fibers prior to the application of an external load. Metallographic studies support the calculations and show that fibers do not fail during the thermal cycling process.

C. Application of Fault Tree Techniques to Failure Analysis of Engineering Materials

The use of fault tree methodology as a logic and graphical aid in the analysis of the failure of engineering structures is a developing discipline. Sufficient use has been made of fault tree methodology to indicate that its techniques may be a valuable asset in the failure analysis of materials. Louthan has applied fault tree techniques in the assessment of the fracture process in high strength low alloy steels (HSLA) (11).

Masters et al have applied fault tree methodology to the failure analysis of fibrous laminated composite systems (12). Their selection of a top event is "physical separation of the laminate into two or more pieces due to quasi-static tensile load." They characterized the occurrences which could precipitate composite failure as delamination, transverse or oblique failure, or a combination of the two and developed the events which lead

to failure. The goal of their analysis was to develop the fault tree and determine the independent sets of basic events, called minimal cut sets, which could precipitate composite failure. They identified the chains of basic events and defined areas requiring further investigation.

The task of failure analysis in composite systems becomes less formidable when the analyst has an understanding of fault tree methodology and an awareness of the events which occur in the failure of a composite system. Fault tree techniques and the details of system performance can be combined to determine clearly and distinctly those events, and only those events, required for system failure, whether the events are inherent material properties, specific environmental exposure conditions or distinct combinations of both. The result of the combination of fault tree methodology and failure analysis is a clearer understanding and a more adequate description of the mechanisms involved in system failure.

III. FAULT TREE METHODOLOGY

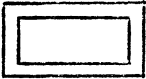

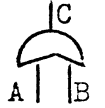

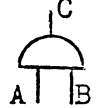
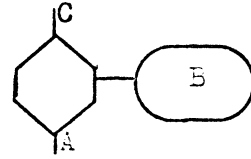


The inherent value of fault tree methodology lies in two important aspects. Fault tree analysis provides a method for acquiring only the specific information required for understanding a given mechanism, and the methodology enables the logical ordering of the information so that a cause and effect sequence is obtained. Fault tree methodology may be divided into four basic phases: system definition, fault tree construction, qualitative evaluation and quantitative assessment. (12) The system can be defined as the combination of all the possible occurrences or events which can be logically related to cause the undesired event. In the analysis treated in this thesis, the top event is failure of a composite material. System events can be either normal or fault events. They can involve operational procedures, operational errors, environmental effects, material properties and possible but unexpected circumstances.

Construction of a fault tree initially involves the selection of the top or undesired event. In failure analysis the undesired event is the one that defines system failure and could be reduction in strength, reduction in ductility, reduction in stiffness, visible cracks of a critical size or separation by fracture. The

top event in this study is separation of a boron-aluminum composite into two or more pieces.

In the construction of a fault tree the top event is designated by a double rectangle (Table 1). Once the top event is selected, the antecedent events, whether normal or fault, are related and subdivided into components until the top event is defined by basic events, which may be either normal or fault events, and by undeveloped events. Fundamental events in a fault tree are called basic events and are events which cannot or need not be divided. Basic events are therefore only causal events and are designated by circles. An event designated by a diamond is an undeveloped event which could be developed further. A double diamond is used if such development is needed before the fault tree is fully understood. Thus diamonds indicate areas where research is needed. Events in the logical sequence between the basic or undeveloped events and the top event are designated by rectangles. A triangle is used to transfer the logical sequence to another part of the fault tree. A list of the symbols for gates and events is shown in Table 1. The events in a working fault tree are related to each other by connecting logic gates. Qualitative evaluation of the fault tree initially involves an examination of the logic gates to insure that the correct one is used. The output event of a gate can be caused by

Table 1. List of Symbols Used in the Fault Tree.

Symbol	Nomenclature	
	Top event	Event which fault tree defines
	Event	An event that can be further developed
	"Or" gate	Logic gate used when either event A or B can lead to event C
	Undeveloped event	An event that will not be further developed but needs to be, in order to use the fault tree completely.
	"And" gate	Logic gate used when both events A and B must occur for event C to occur
	Inhibit gate	Logic gate used when event A will lead to event C if and only if event B is present
	Basic event	An event which cannot be further developed
	Transfer	Transfers to another portion of fault tree

a combination of several events or by the action of one of several events. The two basic gates are the "and" gate and the "or" gate. If several events are necessary to cause an output event but none are sufficient, then an "and" gate is used to connect the events. For example, to cause an event C, if two events, A and B, are required and both are necessary, then C is the output event of the "and" gate and A and B are inputs. Any one event which is the input of an "and" gate is necessary but alone is not sufficient to cause an output event. The use of an "and" gate implies that only when all the input events are present does the output event occur. If an event C is caused either by event A or event B but both events are not necessary, then event C is the output of an "or" gate and A and B are the input events. The input events of an "or" gate are each sufficient to cause an output event. When any one of the input events occurs then the output event takes place. When any event has only one cause and that cause is both necessary and sufficient then no gate is required at all. An inhibit gate is a special case of an "and" gate indicating that an event C will occur as a result of event A if and only if event B is present. The inhibit gate is used to emphasize the importance of some particular step in the failure process.

The next step in qualitative evaluation of the fault tree is to determine the sets of basic events which can independently lead to the top event. Each set of basic events, called a minimal cut set, is a collection of events all of which are necessary and, when taken in total, are sufficient to cause the top event. Depending on the complexity of the fault tree, the minimal cut sets can be determined by inspection, or by computerized techniques using Boolean mathematics and Monte Carlo techniques (12).

The major goal of quantitative evaluation of the fault tree is to calculate the probability of occurrence of the top event (2). Additionally, it is helpful to calculate the importance of each minimal cut set to the top event and the importance of specific basic events to the top event. Evaluation of the minimal cut sets can determine the most probable paths to failure. To accomplish this evaluation a probability of occurrence must be assigned to each basic event and to each undeveloped event in the fault tree. One of several techniques may then be applied for quantification of the fault tree (13, 14, 15).

A simple example demonstrating the use of fault tree analysis for a flashlight that fails to shine is shown in Figures 1 and 2. A simple electrical circuit for the light (Figure 1) shows the bulb, battery, switch and wire. A fault tree to determine the possible events which could

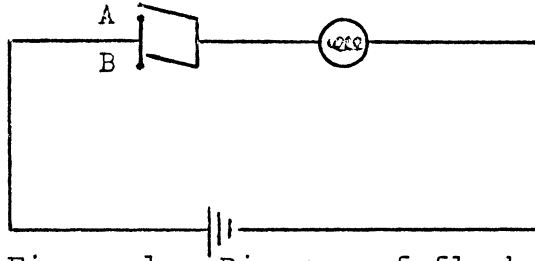


Figure 1. Diagram of flashlight circuit.

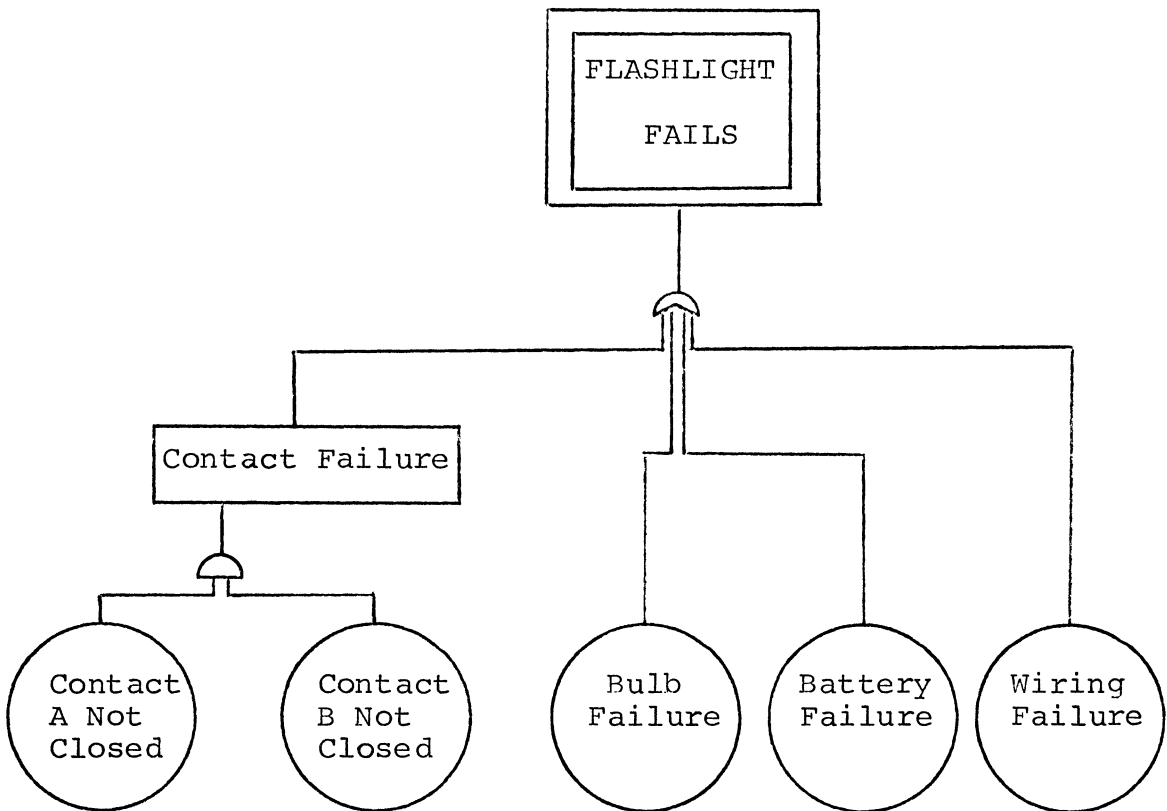


Figure 2. Fault tree for failure of flashlight.

cause the top event ("Flashlight does not shine") is shown in Figure 2. The fault tree construction and qualitative evaluation involves determining the events that could lead to failure and then relating these events in a logical manner.

IV. MATERIALS AND SPECIMENS

The antecedent studies in the fault tree analysis of composite materials which lead to this research considered multi-layered laminates and developed a qualitative fault tree relating basic microscopic mechanisms to macroscopic events such as delamination, transverse fracture and mixed mode failure (12). "Basic" events in that fault tree were events such as load applied, matrix overstressed, fiber overstressed and other steps in a general failure process; however, no attempt was made to relate the events described to the structure and properties of the materials used in a specific composite. When attempts were made to relate the fault tree to failure of boron-aluminum composite the need for a simpler "first step" fault tree was apparent. Furthermore, the necessity for relating the "first step" fault tree to microscopic and macroscopic observations made with failed composites was also obvious. Therefore the research program was divided into two basic-interrelated efforts: (1) Development of a simplified working fault tree, and (2) metallographic studies of failed boron-aluminum composites. Selection of the boron-aluminum system was made after discussions with Dr. D. R. Tenney, NASA Technical Officer for the research program. The composites evaluated were supplied by Dr. Tenney and are

0° unidirectional 8 ply samples pulled to failure in tension at room temperature. The volume fraction of fibers in each sample is approximately 48% and the fiber diameter is 5.6 mil.

The samples examined were part of a NASA experimental program to study the influence of temperature excursions on the tensile properties of boron-6061 aluminum composite laminates. The tensile data is described in Reference 9; pertinent data is reproduced in Table 2. Samples having 6 different thermal histories have been studied. The T4 and T6 treatments represent elevated temperature excursions and are standard anneals for age hardenable aluminum alloys. The T4 anneal is a solid solution treatment and the T6 treatment causes precipitation hardening of the aluminum matrix. Both treatments change matrix properties from the F or as-fabricated condition. Low temperature excursions are represented by liquid nitrogen quench , N, and should have only changed the residual stress state in the composite. Low temperature treatment should not have affected the metallurgical condition of the fibers or induced any interfacial chemical reaction. The liquid nitrogen quench could, however, be expected to cause plastic yielding of the matrix. Descriptions of the thermal excursions are given in detail in Table 3.

TABLE 2

INFLUENCE OF TEMPER CONDITION ON THE TENSILE STRESS-STRAIN
BEHAVIOR OF UNIDIRECTIONAL BORON-ALUMINUM

Temper Condition	E_x (Msi)	ν_{xy}	σ_x^y (ksi)	ϵ_x^y (%)	σ_x^u (ksi)	ϵ_x^u (%)	ϵ_y^u (%)
F	33.0	-*	29	0.094	253	0.886	-0.246
FN	32.9	0.218	69.3	0.214	231	0.783	-0.207
T4	33.1	0.251	43.0	0.133	192	0.671	-0.196
T4N	32.2	0.233	90.6	0.284	162	0.534	-0.124
T6	33.7	0.237	80.4	0.243	185	0.613	-0.145
T6N	33.1	0.213	117.1	0.356	176	0.556	-0.132

TABLE 3

DESCRIPTION OF THERMAL HISTORIES
OF COMPOSITE SPECIMENS

F	As Fabricated.
FN	As Fabricated. Quenched in liquid nitrogen at -320°F for 5 minutes.
T4	Heated at 980°F for 30 minutes. Quenched in water at room temperature.
T4N	Heated at 980°F for 30 minutes. Quenched in water at room temperature. Quenched in liquid nitrogen at -320°F for 5 minutes.
T6	Heated at 980°F for 30 minutes. Quenched in water at room temperature. Annealed at 350°F for 8 hours.
T6N	Heated at 980°F for 30 minutes. Quenched in water at room temperature. Annealed at 350°F for 8 hours. Quenched in liquid nitrogen at -320°F for 5 minutes.

V. EXPERIMENTAL PROCEDURES AND OBSERVATIONS

The initial experimental studies used ordinary metallographic techniques to examine the fracture surface of a failed composite and to examine the fiber damage remote from the site of final fracture. Fractured specimens from the NASA experimental program were prepared for metallographic examination by separating the area of the final fracture from the rest of the sample. A low speed saw with a diamond wafering blade was used to cut the sample. The section containing the final fracture surface was mounted in a bakelite holder with the length of the sample perpendicular to the line of sight. The matrix was abraded with 400 and 600 grit grinding paper. The samples were then polished using 1 μ alumina powder on a standard polishing wheel.

Macroscopic examinations of failed samples in each of the six temper conditions showed that fiber damage remote from the fracture was increased by all thermal treatments (Figures 3-8). Two types of fiber damage were noted: The first was an apparently clean, single break in the fiber with little evidence of separation of the fracture surface (Figure 9a). The second type, however, showed apparent separation of the fracture surfaces (Figure 9b), and microscopic studies indicated disintegration or

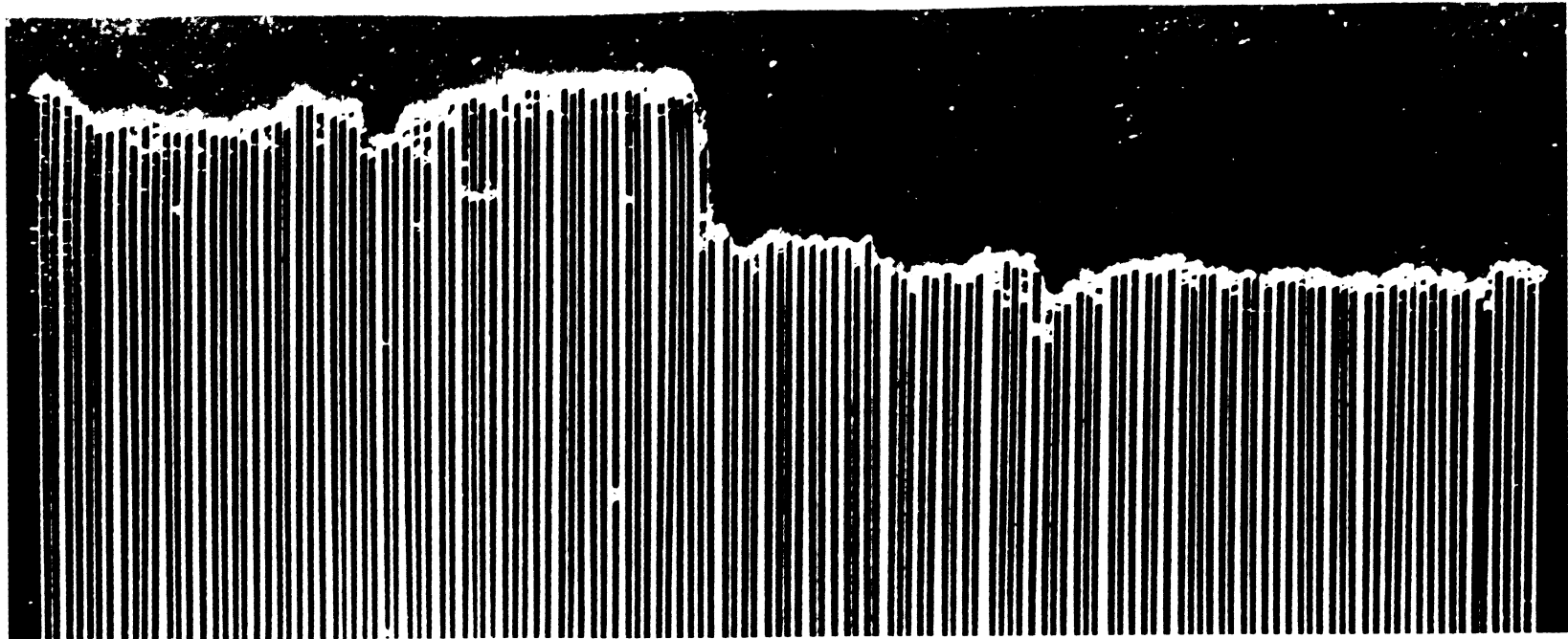


Figure 3. Cross-section of boron-aluminum composite near fracture. F heat treatment.

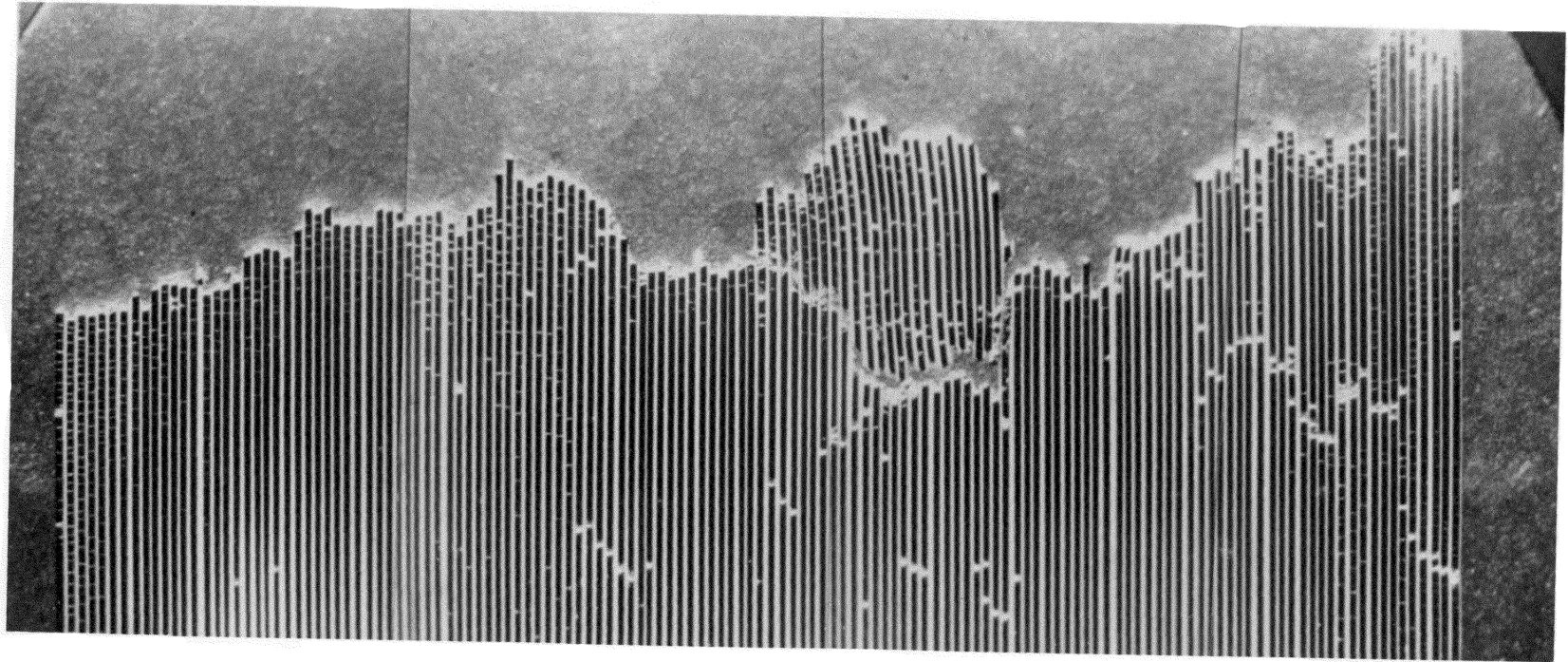


Figure 4. Cross-section of boron-aluminum composite near fracture. FN heat treatment.

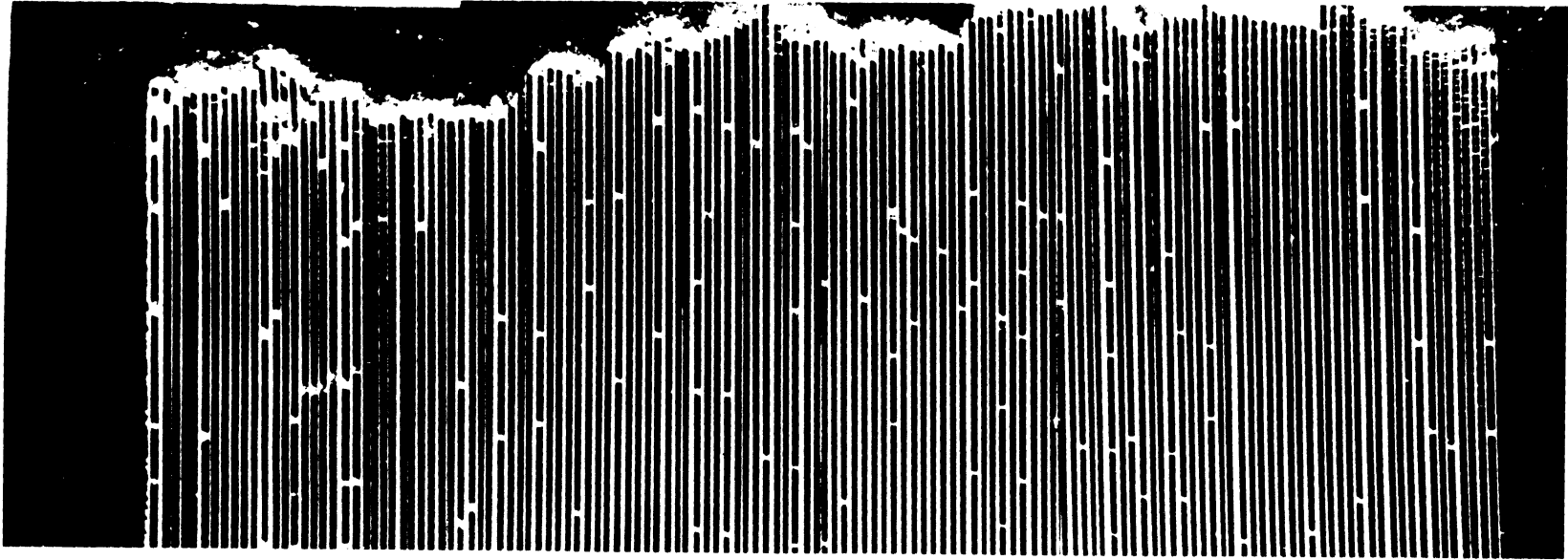


Figure 5. Cross-section of boron-aluminum composite near fracture. T4 heat treatment.

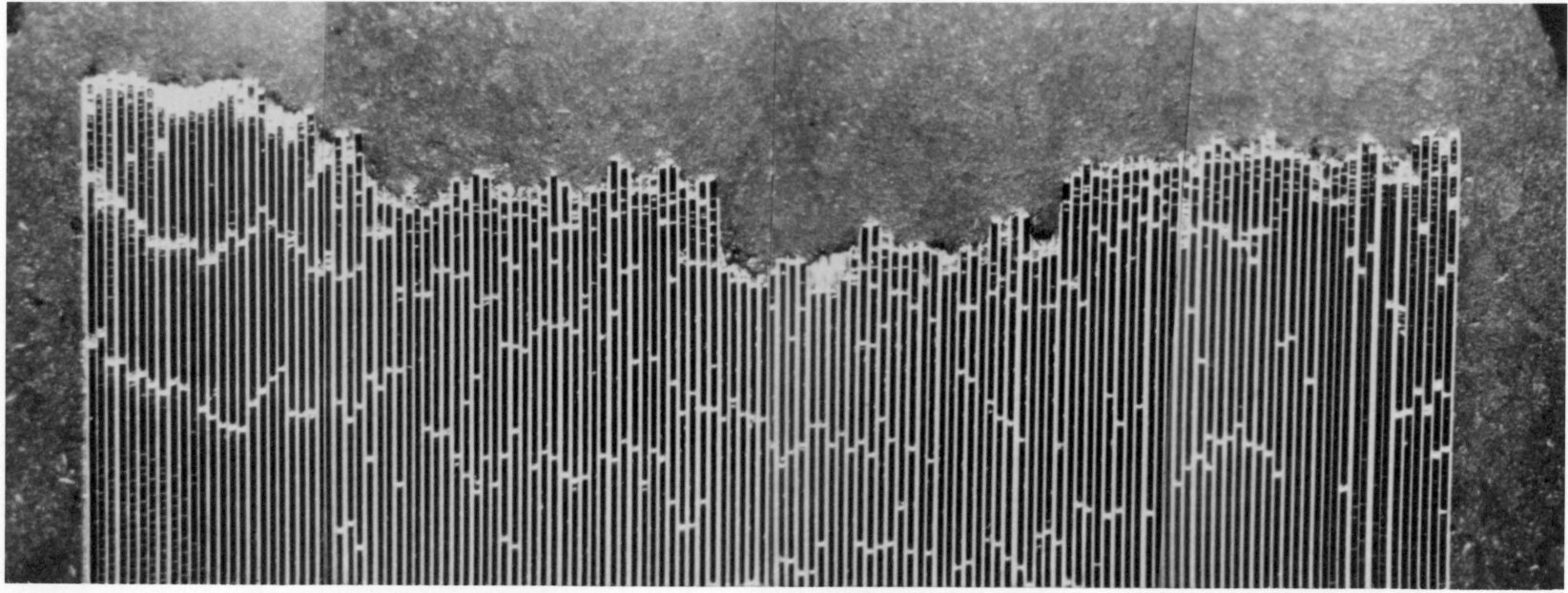


Figure 6. Cross-section of boron-aluminum composite near fracture. T4N heat treatment.

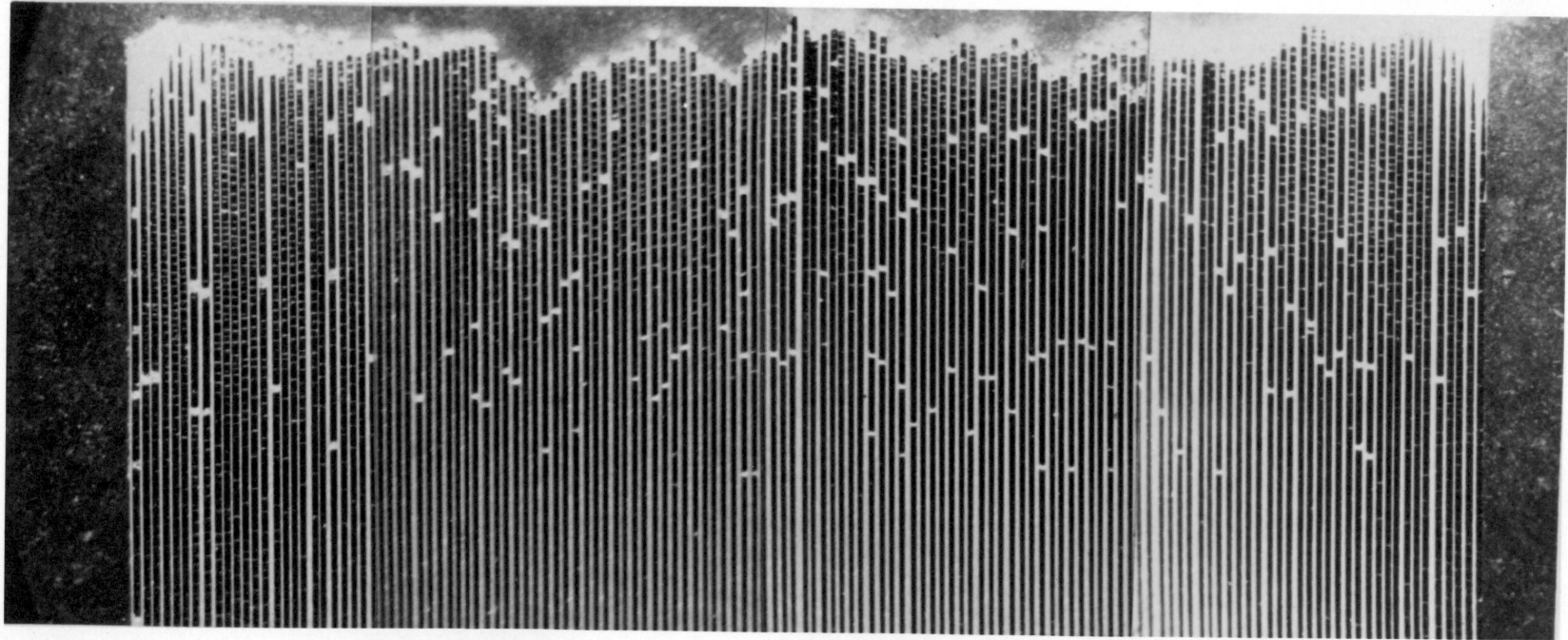


Figure 7. Cross-section of boron-aluminum composite near fracture. T6 heat treatment.

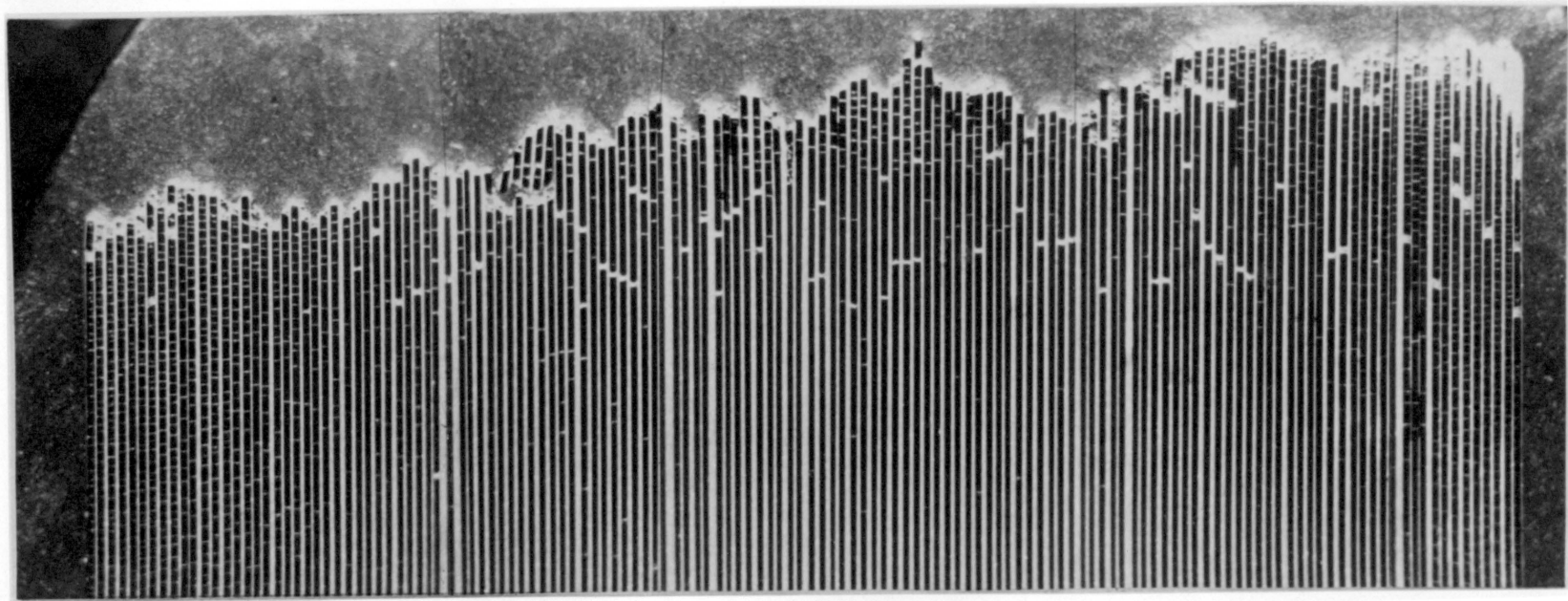


Figure 8. Cross-section of boron-aluminum composite near fracture. T6N heat treatment.



Figure 9a. Cross-section of boron-aluminum composite near fracture showing type I fiber breaks. T6N heat treatment. Magnification 50x.

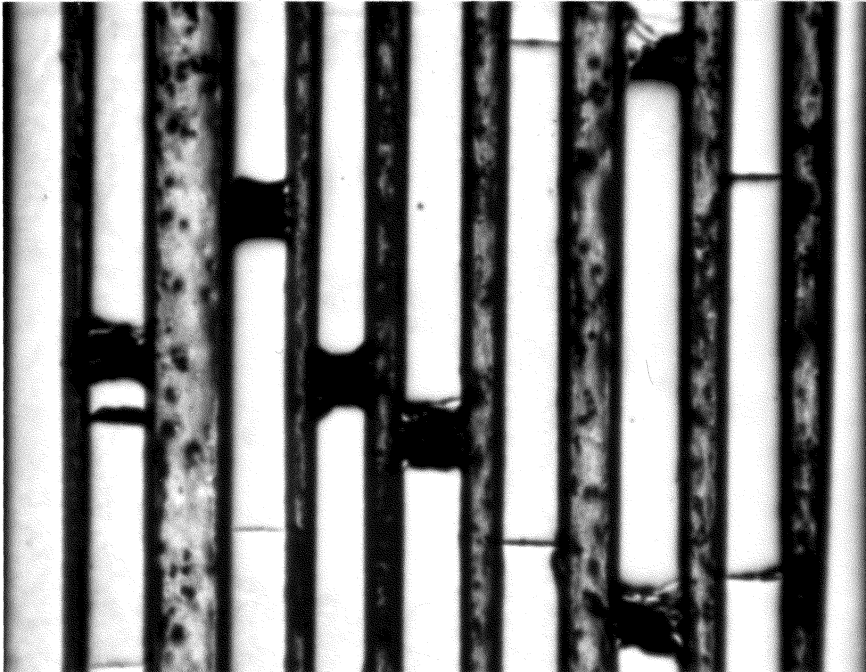


Figure 9b. Cross-section of boron-aluminum composite showing type II fiber breaks. T6N heat treatment. Magnification 50x.

multiple cracking of the fiber ends. Fiber damage of both types was seen in all samples regardless of heat treatment. Examination of a single fiber shows that where breaks of the second type occur, breaks of the first type are not generally evident.

In two of the three temper conditions the liquid nitrogen treatments increased the tendency for breaks of the second type. The increases were seen in the FN and T4N heat treatments. However the reverse was true for the precipitation hardened samples and fewer type II breaks were evident in the T6N than in the T6 heat treatments.

Quantitative microscopy of failed composites has established a relationship between the mean fiber length and the distance from the fracture surface. It was found that the mean fiber length increased with distance from the fracture and that the temper condition was a factor in determining the rate of increase. In each of the temper conditions studied it was found that the mean length of fiber at the fracture was 650 microns (Figure 10). This mean fiber length could be related to the ineffective length described by Rosen (16). The ineffective length is the length where the fiber loses its load bearing and load transferring characteristics.

Subsequent to the preparation of these initial photomicrographs (Figures 3-9), some concern developed regarding the effects of mechanical abrasion techniques on the integrity of the material being prepared for examination. It was felt that once the brittle boron fibers were exposed and free from restraint by the aluminum matrix that the boron may be broken by the grinding involved in specimen preparation. As long as care was taken and the fibers only slightly exposed no degradation of the fibers by abrasion was noted. However, if an attempt was made to reduce specimen preparation times by using grinding paper with coarse grit, or if it was desirable to expose more of the fiber for analysis, the fibers tended to crack perpendicular to their axis with a single clean break (The type I break). Thus specimen preparation techniques are critical in obtaining accurate evaluations of composite failure.

In an attempt to reduce the possibility of attaining misleading results because of mechanical specimen preparation techniques, a second approach was developed. This approach involved the use of chemical techniques to preferentially leach away the aluminum matrix thus exposing the boron fiber without the mechanical action. Initially this approach was used to determine the correlation between the fracture morphology in the mechanically prepared and

the chemically prepared specimens. In each case where type II breaks in the chemically prepared specimen was compared to the type II breaks in the mechanically prepared specimen there was a one to one correspondence in the number of breaks and in the fracture morphologies. The nodular microstructure of the boron fiber and its resistance to chemical attack by the corrosive media made it difficult to characterize the correspondence between type I breaks in the chemically and mechanically prepared specimens. Additional metallographic studies, carefully conducted and involving only mechanical polishing of the specimens without the use of abrasive paper, indicated that type I breaks of the boron fiber were indeed characteristic of fracture. Subsequent scanning electron microscopy revealed the appearance of type I breaks in the final fracture surface. These breaks are characterized by a fiber fracture surface which is smooth and free of multiple fragments (Figure 19).

The chemical techniques were also used to leach away the aluminum matrix material, thus exposing the intact boron fibers. The chemical reagent used in the leaching technique was 4.5 molal solution of commercial grade hydrofluoric acid in distilled water. The samples were placed in the solution and the aluminum attacked chemically until the fibers were exposed. The boron fibers were noted to

have the nodular microstructure typical of the high strength amorphous boron (17). The dissolution process could be discontinued at any time by removing the sample from the solution. The optical microscope was then used to determine if the fibers were sufficiently exposed for analysis.

The acid solution tended to preferentially attack locally the area of the matrix next to a broken fiber (Figure 11). The surface lamina shown in Figure 11 is from a sample in the T4N temper condition which has been chemically reacted to expose the boron fibers. Each of the dark spots in the matrix material adjacent to a fiber is an area in the matrix of preferred chemical attack. This localized chemical attack of the matrix at the break site is characteristic of all temper conditions and is shown at a higher magnification in Figure 12. A broken boron fiber is seen in the area of greatest matrix dissolution.

The localized attack suggests two complimentary reactions which aid the dissolution process. The breaking of a fiber is likely to cause separation of fiber fragments from the matrix material. This separation exposes new surface area in the matrix to the acid solution and consequently dissolution of the matrix occurs at a higher rate at the fiber break site relative to the bulk of the matrix. Additionally the breaking of the fiber may induce some local strain in the matrix at the break site. Thus the

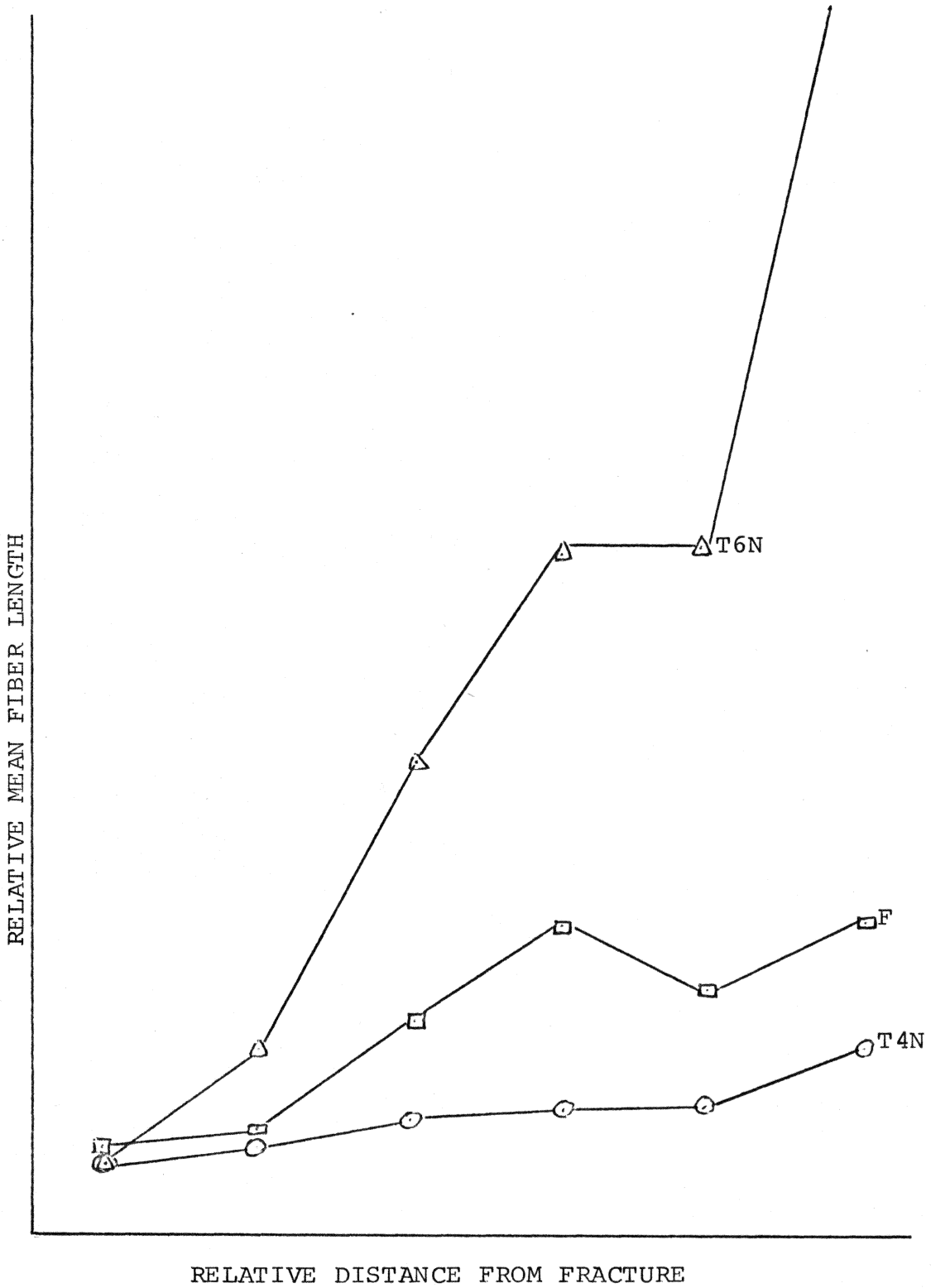


Figure 10. Relationship between mean fiber length and distance from the fracture surface.

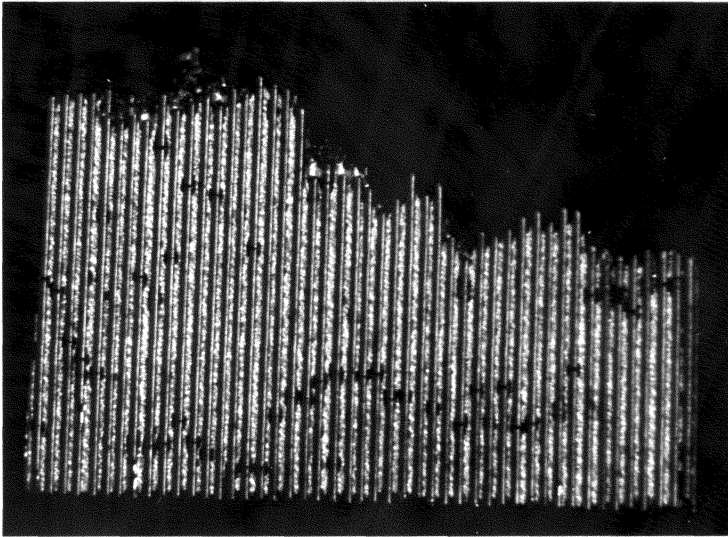


Figure 11. Surface lamina in the boron-aluminum system showing localized dissolution of material at break sites. T4N heat treatment. Magnification 10x.

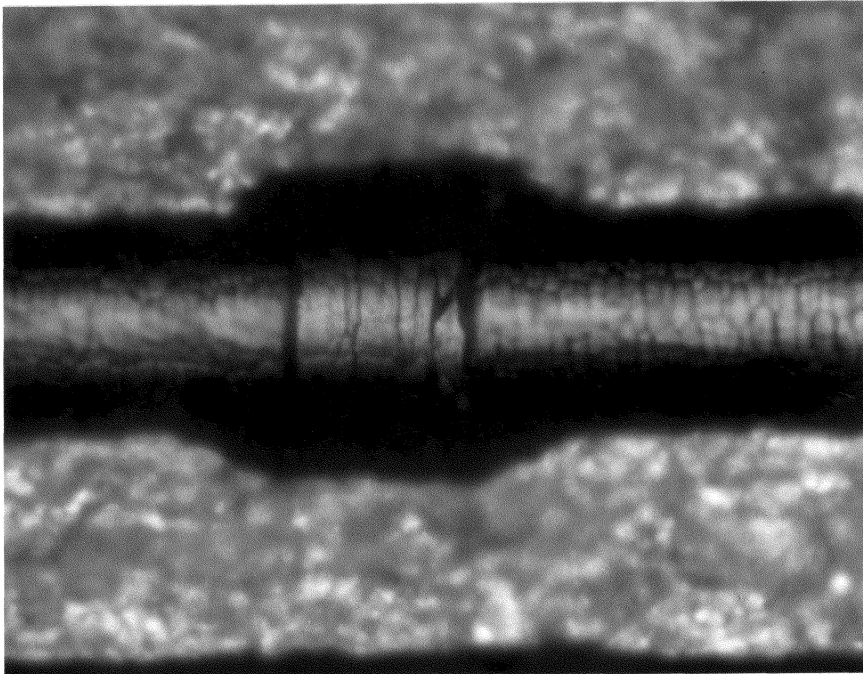


Figure 12. Exposed boron fiber showing multiple cracking of fiber and localized dissolution of matrix material. FN heat treatment. Magnification 100x.

break site becomes anodic relative to the rest of the matrix. Because localized cold work makes a material only slightly more anodic relative to an annealed piece of the same material and because the entire matrix may have already experienced some cold work due to thermal treatment, it is likely that new surface area at matrix-fiber separation is the phenomena of greatest significance.

An attempt was made to determine if the fiber fracture morphology found in one lamina was also characteristic of the fiber fracture morphology between the lamina. The metallographic studies conducted (Figures 3-8) and an examination of the literature indicate that in the boron-aluminum composite system fibers do not break randomly following the onset of fiber degradation. The fiber degradation process is zone oriented. Fiber breaks tend to localize with many breaks in one fiber and with a number of adjacent fibers broken. The results of the chemical leaching studies indicate that where the bonding between the lamina is sufficient to prevent delamination during tensile failure, fiber breaks tend to propagate between the lamina by a damage zone process. A surface lamina with several broken fibers is shown in Figure 13. There is a distance between the fiber ends because the acid attack has dissolved the matrix locally and the fiber fragments held by the matrix have been removed.

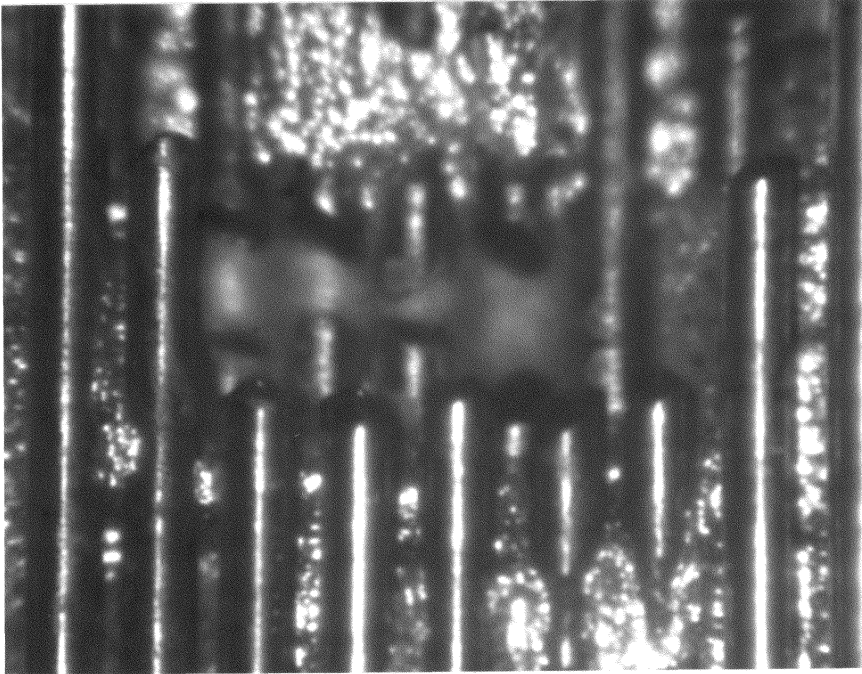


Figure 13. Surface lamina of boron-aluminum composite showing relationship in fiber breaks between lamina. T4N heat treatment. Magnification 50x.

The absence of the fragments and dissolution of the underlying matrix has exposed the fibers in the next lamina. Shown in this layer are additional fibers with the same fracture characteristics as those in the surface layer.

It has been found that if the bonding between lamina is not sufficiently adequate to prevent delamination that the above phenomena does not occur. In the investigation of a specimen with a T4N thermal treatment, the matrix material was leached away to expose a fracture morphology of the type noted in Figure 13. When the chemical attack reached the state of advancement indicated in Figure 13, there were no fiber breaks seen in the underlying lamina which could be directly related to fiber breaks in the surface lamina. As the acid attack was continued the entire first lamina in the composite section being investigated came off intact with sufficient matrix material between the fibers to hold the lamina together.

Fracture surfaces of samples of each of the six thermal treatments were also studied by scanning electron microscopy to determine if changes in the fracture morphology resulted from the thermal treatment. Emphasis was placed on determining if as a result of thermal treatment (1) there was any notable change in the fracture characteristics of the matrix, (2) if type one and type two fiber breaks were obvious in the fracture surface, (3) if the fibers ex-

perienced sufficient localized debonding to pull out of the matrix exposing a length above the fracture surface, and (4) if there was any visible chemical reaction zone at the fiber-matrix interface.

Examination of samples from each of the six thermal treatments indicates that regardless of the type and degree of the thermal excursion the matrix material fails by microvoid coalescence. A typical aluminum fracture surface indicative of all of the specimens examined is shown in Figure 14a. Boron fibers are seen in the bottom and upper left and right hand corners of the photomicrograph. A shear lip was formed in the matrix material as it pulled away from the fiber in the final failure process. This shear lip can be seen in the bottom of Figure 14a adjacent to the boron fiber. Figure 14b is a higher magnification of the same area showing the extremely ductile nature of the matrix. There was no visible difference in the fracture morphology in the matrix as the thermal treatment was changed except perhaps for some changes in the size of the microvoids as can be seen by comparing figure 14a to figure 15.

Examination of the fracture surface of the specimens by scanning electron microscopy indicates that the bonding between the lamina was effective and in general there is no evidence of debonding between lamina as a characteristic

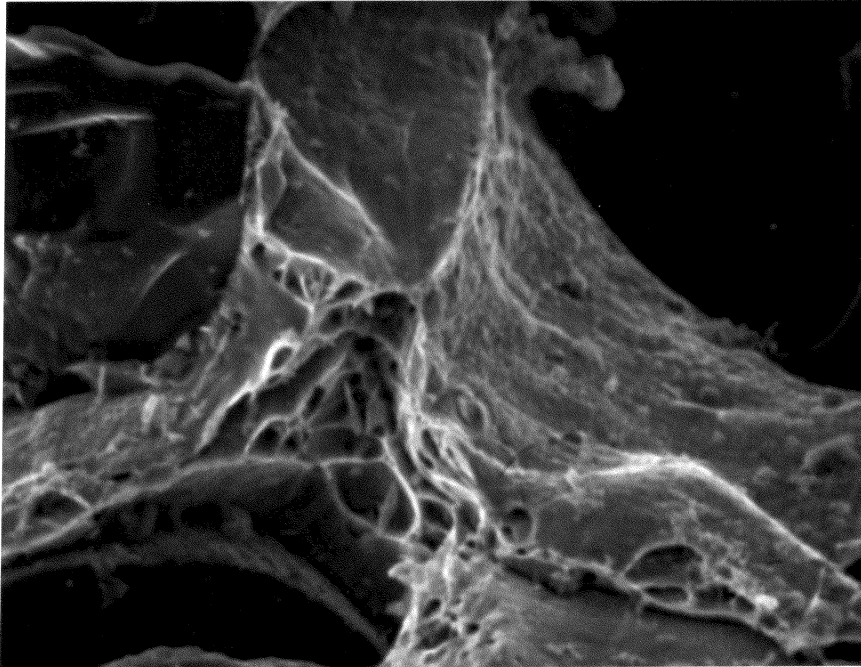


Figure 14a. Fracture surface of boron-aluminum composite showing ductile failure of the aluminum matrix. T4 heat treatment. Magnification 500x.

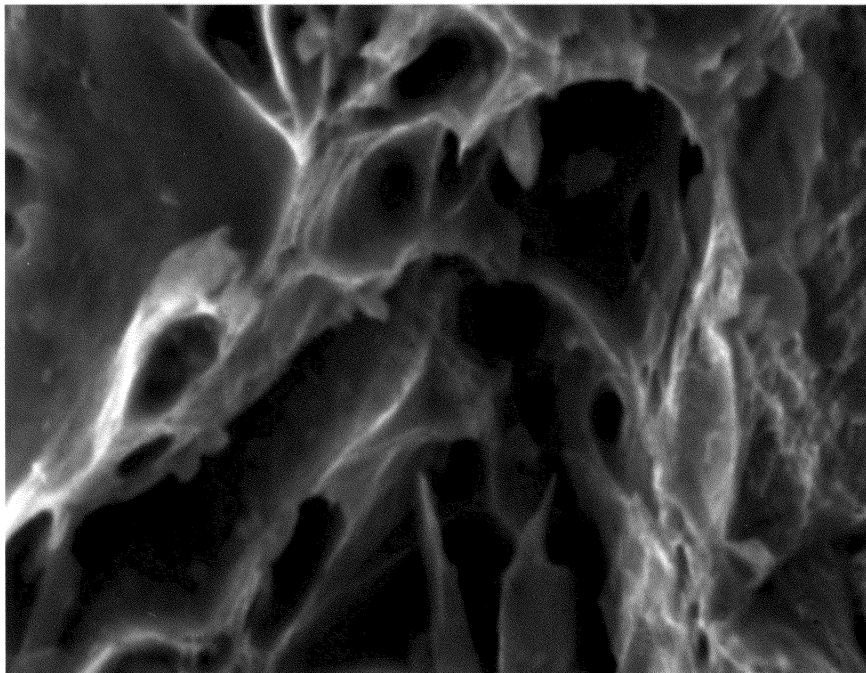


Figure 14b. Fracture surface of boron-aluminum composite showing microvoid coalescence in the aluminum matrix. T4 heat treatment. Magnification 2000x.

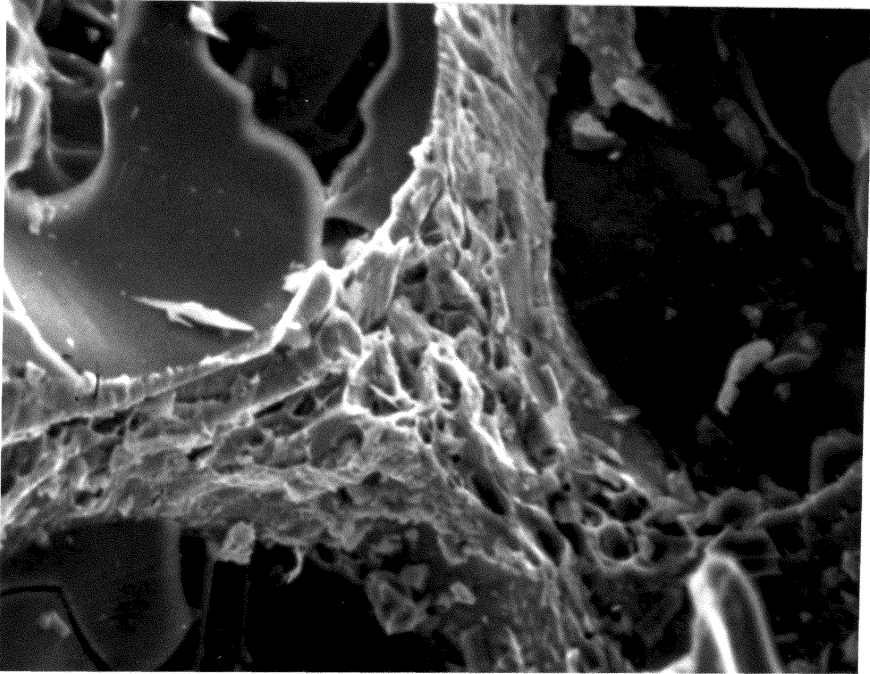


Figure 15. Fracture surface of boron-aluminum composite showing microvoid coalescence in the aluminum matrix. T6N heat treatment. Magnification 500x.

of fracture. However it can be seen in figure 16a that for the specimen with the T6 thermal treatment there is some ductile tearing between the lamina during the failure process. Figure 16b shows this phenomenon at a higher magnification. The phenomenon was also present in the specimen with the T4N heat treatment but the ductile tearing was noted to be much less severe and more localized for the T4N treatment (Figure 17). Specimens with other thermal histories exhibited little or no evidence of debonding between lamina in the area of final fracture. A photomicrograph of a fracture surface in which interlaminar debonding is not evident is shown in figure 18.

It was also noted that in all the samples studied the two types of fiber fracture were apparent: a clean, single break in the fiber with no separation of the fiber ends (Figure 9a) and the second type of break with a separation of the fiber ends at the break site with the fiber break exhibiting multiple cracking (Figure 9b). A type I break in the final fracture surface of the material as revealed by SEM is shown in Figure 19. Although the number of type I fiber fractures seen on the failure surface of any one sample is small relative to the number of multiply cracked fibers, type I failures appear in fracture surfaces of samples from all heat treatments. A type I fiber break surrounded by type II fiber breaks is

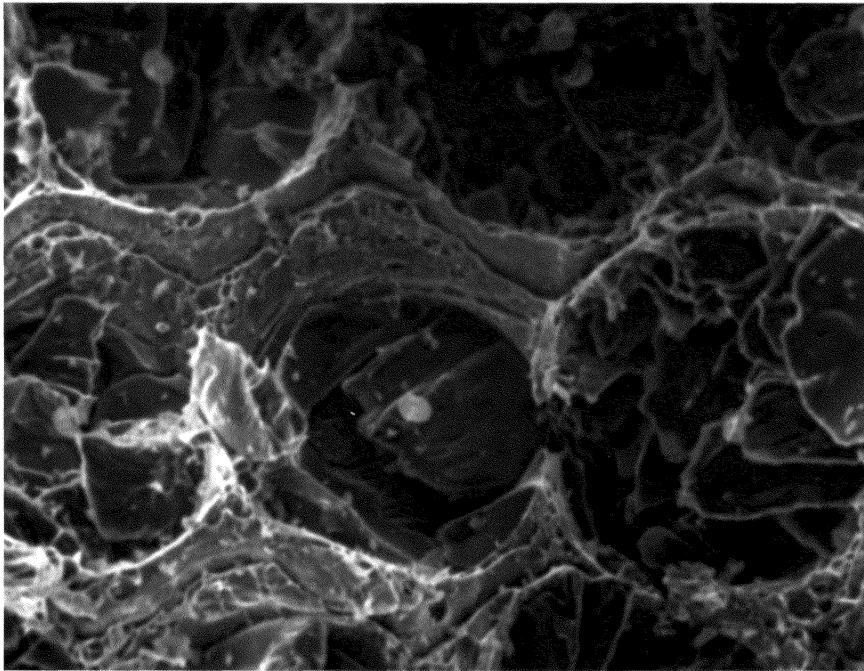


Figure 16a. Fracture surface of boron-aluminum composite showing separation of the lamina. T6 heat treatment. Magnification 200x.

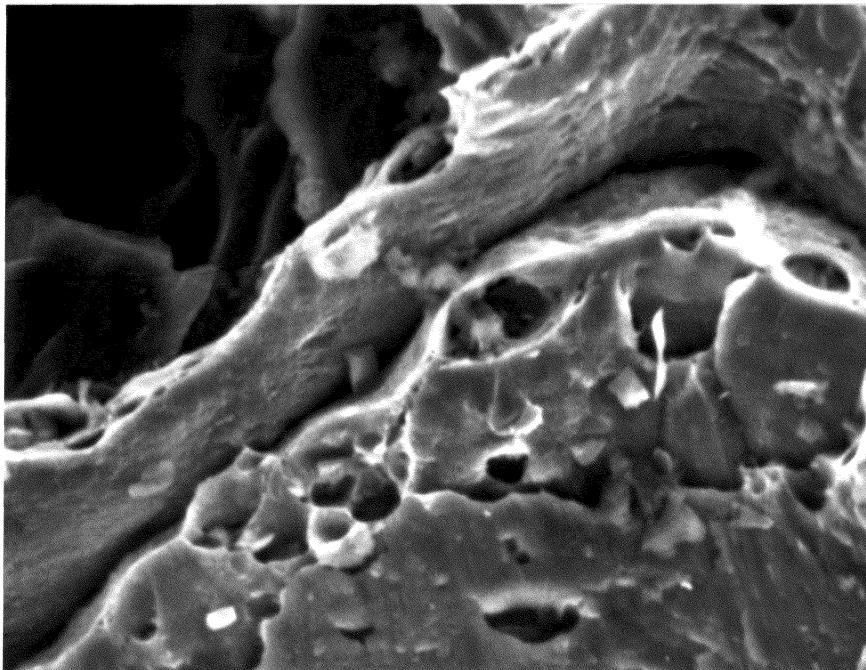


Figure 16b. Fracture surface of boron-aluminum composite showing separation of the lamina. T6 heat treatment. Magnification 1000x.

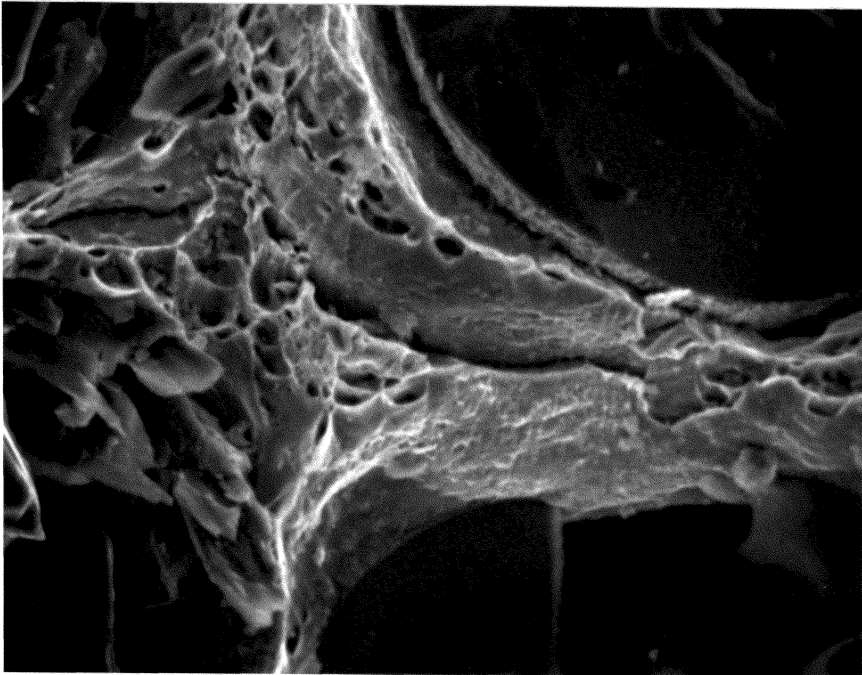


Figure 17. Fracture surface of boron-aluminum composite showing separation of the lamina. T4N heat treatment. Magnification 500x.

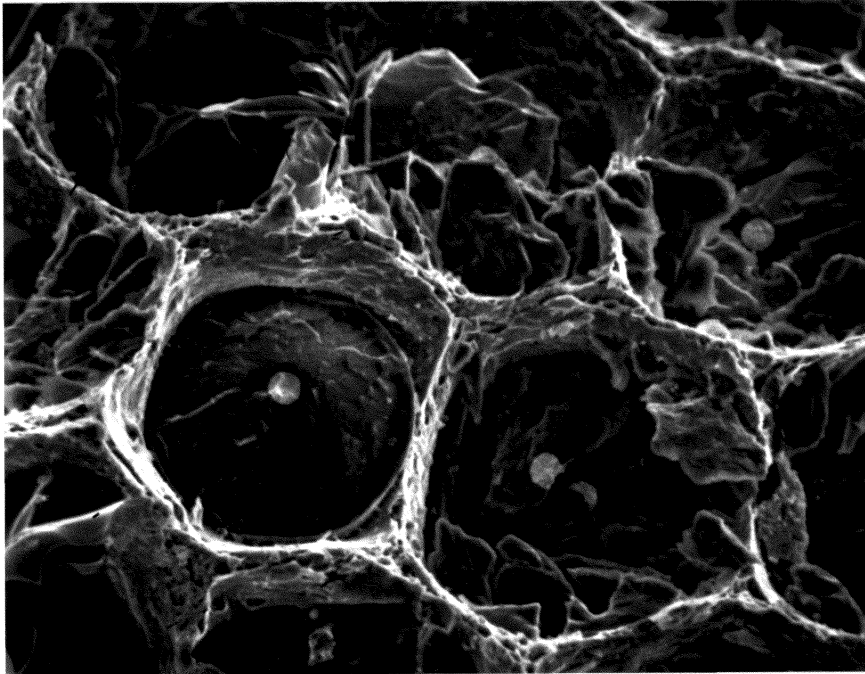


Figure 18. Fracture surface of boron-aluminum composite showing effects of good bonding between lamina. Shown also is a type I fiber break surrounded by type II fiber breaks. F heat treatment. Magnification 200x.

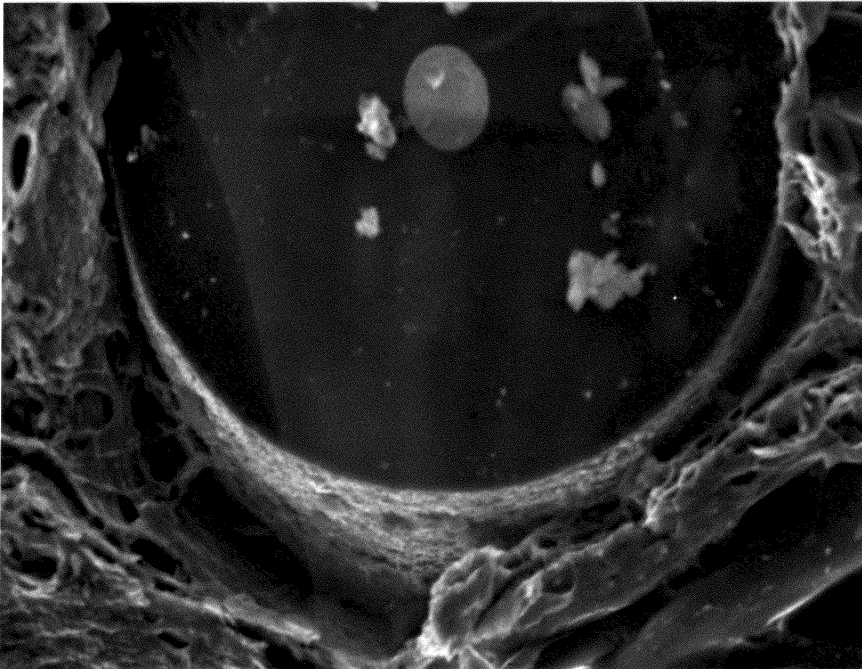


Figure 19. Fracture surface of boron-aluminum composite showing ductile matrix tearing away from the fiber. F heat treatment. Magnification 500x.

shown in Figure 18. The severely damaged and cracked fibers shown in Figure 18 are the most representative of the fiber fracture morphology at the final fracture surface in all heat treatments. These observations indicate that final fracture occurs by coalescence and type II fractures.

In the fracture of fibrous composite materials debonding between fibers and matrix can be seen in the final fracture surface as fiber pull out. The phenomenon occurs when a fiber fails slightly remote from the final fracture surface, possibly within an ineffective length, and due to debonding between the fiber and matrix the fiber pulls out of the matrix and a length of fiber remains above the final fracture surface. Specimens which show the most extensive fiber pull out have been heat treated to the FN and the T4 condition. Fiber pull out in a specimen treated to the FN condition is shown in Figure 20a and Figure 20b shows similar fracture in a sample treated to the T4 condition. Figures 21a and 21b show the same phenomenon from a different sample orientation and at several magnifications for the FN thermal treatment. However, extensive fiber pullout was not noted in any of the specimens.

Although there are cases where the sides of the fiber are exposed, the exposure is not the result of fiber pull

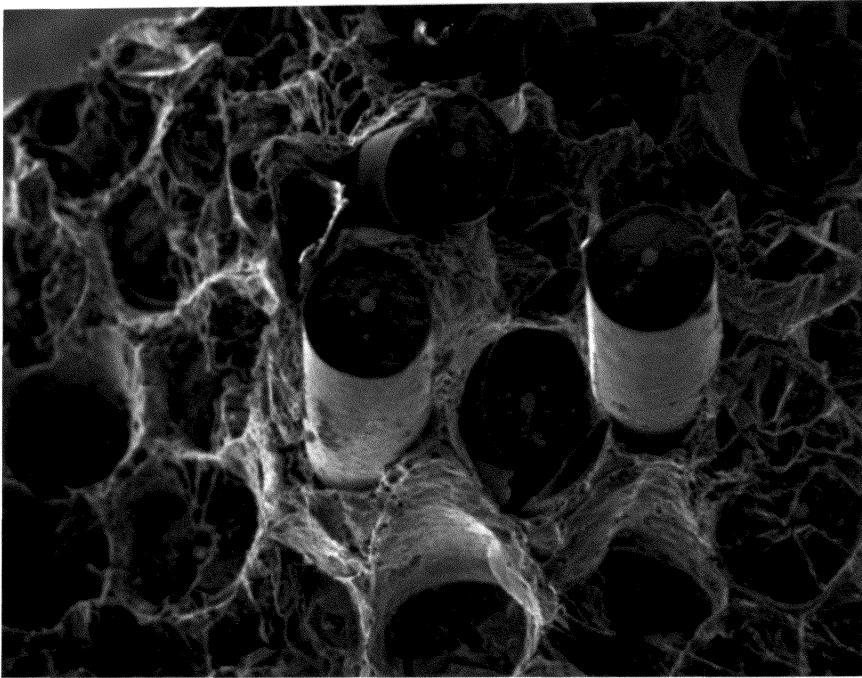


Figure 20a. Fracture surface of boron-aluminum composite showing fiber pullout. FN heat treatment. Magnification 100x.

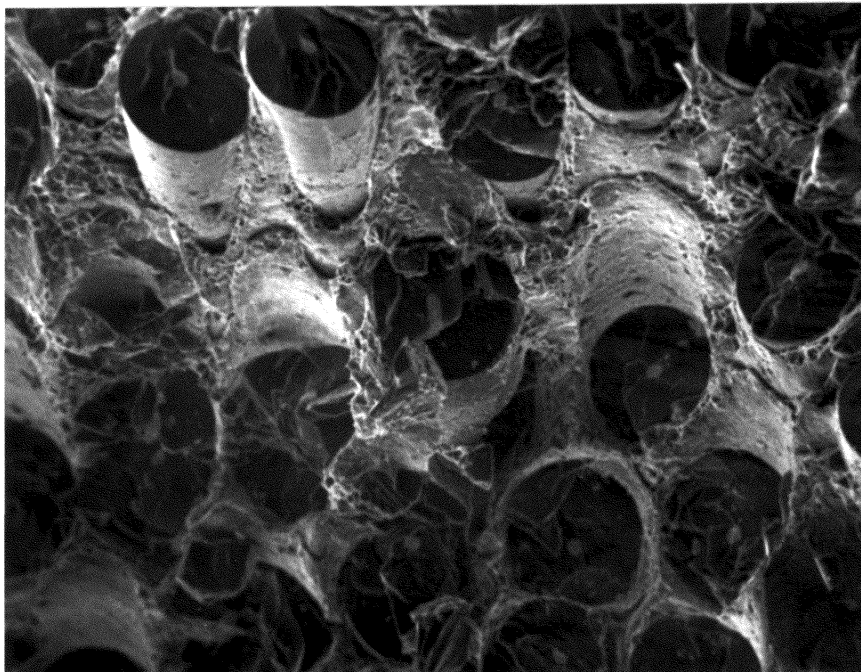


Figure 20b. Fracture surface of boron-aluminum composite showing fiber pullout. T4 heat treatment. Magnification 100x.

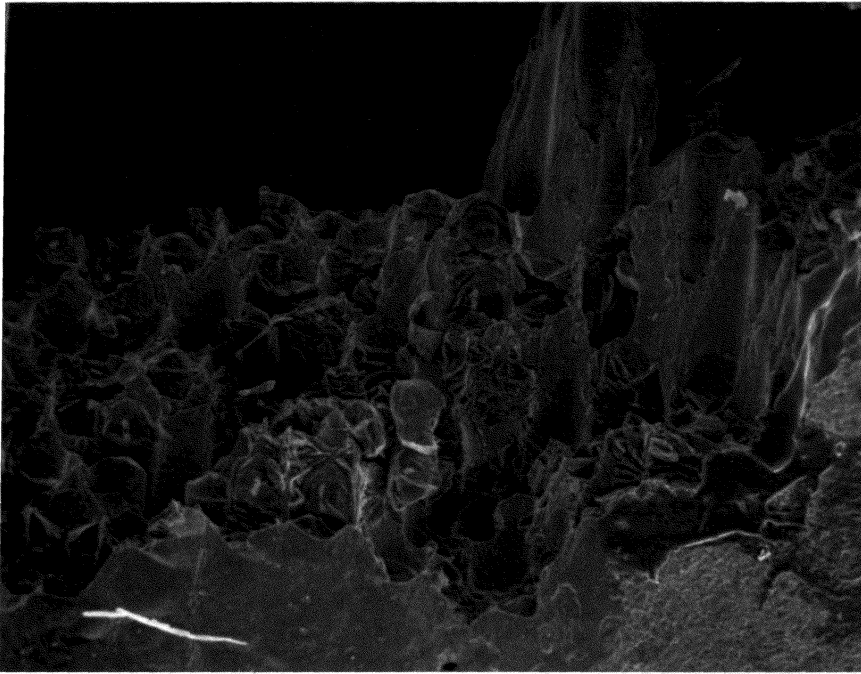


Figure 21a. Fracture surface of boron-aluminum composite showing fiber pullout. FN heat treatment. Magnification 45x.

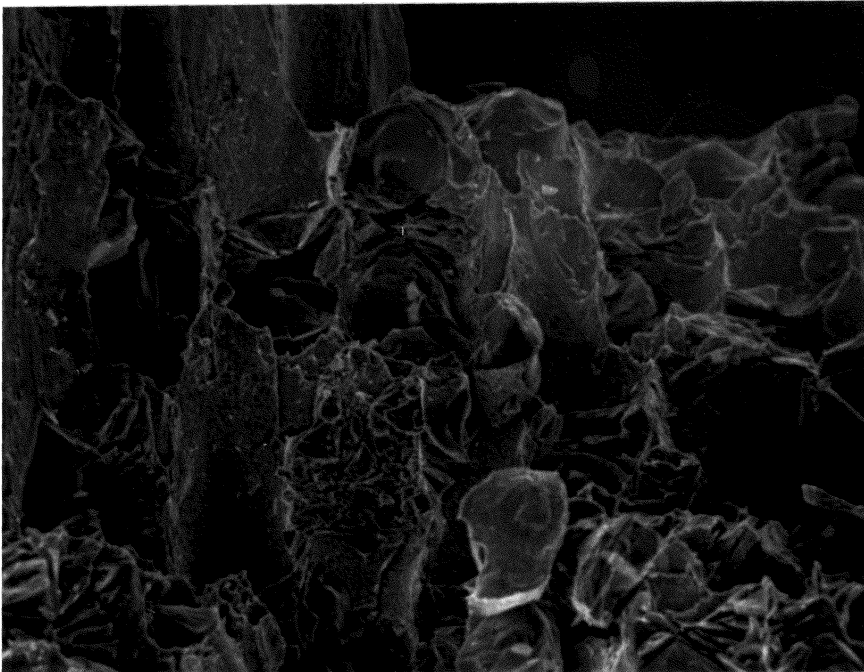


Figure 21b. Fracture surface of boron-aluminum composite showing fiber pullout. FN heat treatment. Magnification 90x.

out. In Figure 14a the sides of the fibers in the bottom of the photomicrograph are apparent. This phenomenon is thought to be due to the matrix tearing away from the fiber as the matrix fails by ductile mechanisms. Figure 19 also shows the same phenomena at a higher magnification.

It was not possible to determine quantitatively with the aid of the scanning electron microscope the nature of the chemical reaction at the interface between the fibers and matrix. Although the formation of the distinct phase during thermal processing is likely from the standpoint of thermodynamics, the rate of such a formation is kinetically sluggish and requires long time treatments at high temperatures and pressures (10). In only one instance in the specimens studied to date has there been extensive evidence of a chemical reaction. This attack produced a visible, additional phase at the fiber matrix interface. A photomicrograph of the fracture surface of a failed boron-aluminum sample in the T4 temper condition and exhibiting a surface zone is shown in Figure 22a. In the upper right corner of the micrograph can be seen a fiber with an unusual fracture morphology. The fracture surface of this fiber is quite flat except for the radial crack which is characteristic of all boron fibers (17). At the edge of the fiber is a lip of a finite thickness which appears to be a distinct interface between the fiber and

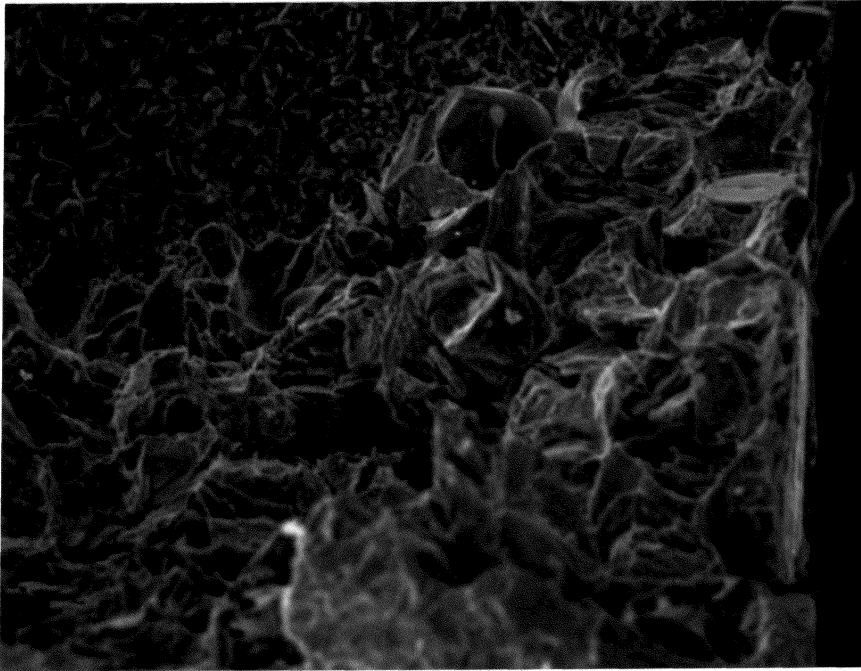


Figure 22a. Fracture surface of boron-aluminum composite showing unique fiber fracture morphology. T4 heat treatment. Magnification 100x.

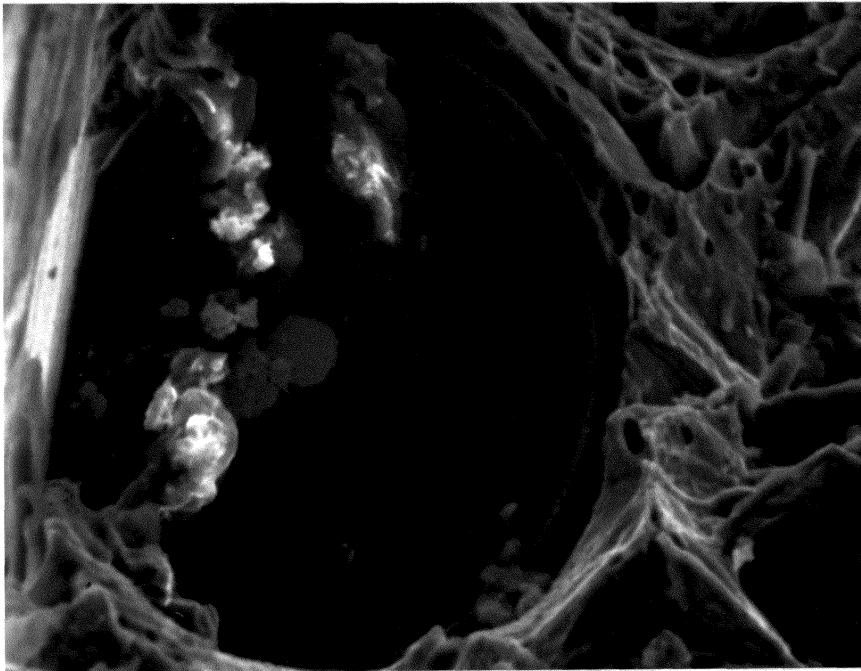


Figure 22b. Fracture surface of boron-aluminum composite showing unique fiber fracture morphology. T4 heat treatment. Magnification 500x.

the matrix material. A photomicrograph of this fiber from a different orientation is shown in Figure 22b. Although the extent and nature of the chemical reaction is not quantitatively known, EDAX analysis indicates that aluminum is present in the interfacial region. EDAX analysis also indicates that there is no silicon or other identifiable elements present in this region.

VI. DISCUSSION

The microstructural phenomena described in the experimental procedures and observations section can be interpreted as showing that previous thermal treatment has a significant effect on the mechanical performance of the composite material. The nature of the changes induced in the material as a result of thermal treatment can be both chemical and mechanical. It is believed that a complex interaction of both chemical and mechanical changes induce matrix-fiber interactions which are responsible for the degradation of the mechanical properties in 0° continuous fiber reinforced composite materials. The data shown in Table 2 indicate that a reduction in ultimate tensile strength (σ_x^u) occurs in all thermal treatments relative to the F or "as fabricated" condition. There is also a corresponding decrease in the strain to fracture (ϵ_x^u) in the thermally treated samples. However, an improvement in the nominal yield strength (σ_x^y) of the material from the F condition is also noted. Because the boron fiber is extremely brittle and can be considered elastic to failure, changes in the nominal yield strength of the composite can be attributed to changes in the mechanical strength of the matrix. The high temperature excursions, T4 and T6, represent a solid solution treatment and a

precipitation hardening treatment respectively for the 6061 aluminum alloy. It is expected that the T4 solid solution treatment will improve the yield strength of the matrix over the as fabricated condition because the as fabricated material is assumed to be in the overaged condition. Additionally the T6 heat treatment which involves the precipitation of a coherent second phase in the matrix would further increase the strength of the matrix. A corresponding increase in yield strength of the composite as a result of this treatment is noted. The lowering of the ultimate tensile strength as a result of these thermal treatments is an apparent anomaly that will be explained. The effect of the thermal treatment at low temperatures, involving a nitrogen quench, is to lower the ultimate strength of the composite and to raise the apparent yield strength. Because of the differences in the coefficients of linear thermal expansion, α_m and α_f , between the matrix and fibers any drastic change in temperature is likely to induce some matrix yielding provided that bonding is good at the interface between the fibers and matrix. The improvement in nominal yield strength in the nitrogen quenched samples is the result of a strain hardening mechanism in the matrix. The effects of thermal treatment on the apparent yield strength and ultimate strength of the composite material are shown graphically in figure 23.

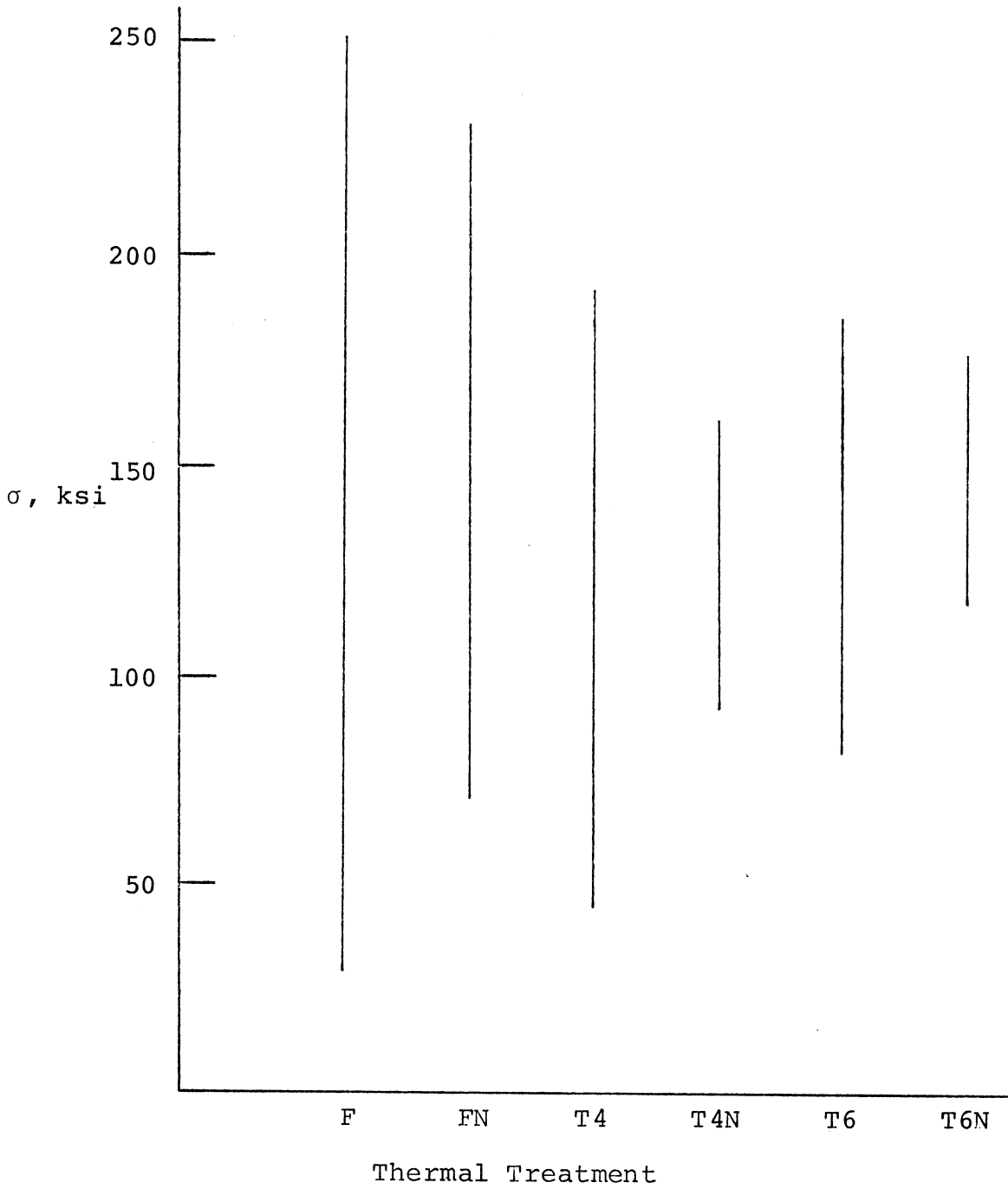


Figure 23. Plastic strain region of various heat-treated boron-aluminum composites.

Changes in the properties of the composite have an effect on the fracture morphology. The effect of any thermal treatment is to increase the amount of fiber damage remote from the final fracture surface. Increases in both type I and type II breaks as a result of increases in the severity of the thermal treatment can be seen in figures 3-8. It is seen in these photomicrographs that when type I breaks occur in a fiber, type II breaks are not generally nearby in the same fiber. This phenomenon is particularly obvious in figure 6. This may indicate that the two types of break occur at different times in the failure process. Possibly type I fiber breaks initiate early in the failure sequence and extensive fiber damage with many fragments is restrained because there is little yielding of the composite associated with isolated breaks. As the process continues and more fiber lengths become ineffective in localized areas, more of the load is transferred to the matrix. Localized yielding of the matrix is then accompanied by the appearance of type II breaks and final fracture of the composite occurs by coalescence of type II breaks.

The effect of the nitrogen quench on the fracture morphology of the F and T4 temper conditions is to increase the tendency for type II breaks. Because the N treatment should not involve any chemical reactions or diffusion

between fibers and matrix, the increase in type II breaks must be attributed to changes in local residual stresses. The N treatment should increase the tensile stresses in the fiber and compressive stresses in the matrix as the specimen is heated from 77°K (liquid nitrogen temperature) to room temperature. Because the increase in residual tensile stress in the fiber is high, fiber fracture occurs at lower stresses than predicted. The T4N thermal treatment is apparently much more severe than the FN treatment and induces a greater number of fiber breaks and failure occurs at a lower stress.

Of the temper conditions exposed to the nitrogen quench, the T6 condition which represents the highest matrix strength was least affected. Comparison of figures 7 and 8 indicate that fewer type II breaks were evident in the T6N sample than in the T6 sample. It would seem that the T6N heat treatment is the most severe thermal treatment because it involves an additional heating and cooling cycle and time for interfacial reactions prior to the nitrogen quench. However, as seen in table 1, the T4N heat treatment involves the most severe degradation in composite properties. This apparent anomaly can be rationalized by considering that the T4N thermal treatment involves a single thermal cycle involving both a high and low temperature. The matrix at the high tem-

perature and prior to the water quench is in a fully relaxed state corresponding to an annealed condition. The quench from heat treating temperature to room temperature and subsequently to liquid nitrogen temperature plastically strains the matrix. If the matrix is assumed to remain fully elastic, the stress in the matrix in the longitudinal direction is calculated to exceed 100 ksi. However, the matrix does not strain only elastically and experiences plastic deformation at stresses well below 100 ksi. The deformation in the matrix is probably not homogeneous and stresses are concentrated locally along the length of the fibers. These local stresses can lead to early failure of the fiber and subsequent composite fracture. The precipitation treatment temperature for the T6N samples is lower than the temperature for the solution treatment. Thus, if the matrix is assumed to be free of residual stresses at this lower temperature of treatment, the fiber stresses generated by cooling to liquid nitrogen temperatures are less in the T6N than in the T4N samples, and the magnitude of the stresses along the fiber length decreases. When loading from an external source commences, the fibers in the T6N material are less susceptible to degradation than those fibers in the samples with the T4N heat treatment. The graph of the mean fiber length versus distance from the fracture surface for the

F, T4N and T6N temper conditions as shown in figure 10 indicates that fewer breaks are found per unit length in the T6N condition. Although the thermal treatment does have an effect on the number of fiber breaks, there is no effect from heat treatment on the mean fiber length at the site of final fracture. Regardless of heat treatment, there is a critical fiber length at which the fiber can no longer accommodate the tensile stress due to the loading condition and at which the matrix loses its ability to transfer the load to adjacent fibers.

Debonding between lamina in the failure process generally occurs in composites where the lamina have lay-ups with several fibers orientations. Debonding occurs due to differences in stiffness and in the coefficient of linear thermal expansion relative to the direction in which the tensile test is conducted. However some debonding was found in the 0° orientation composite samples. Apparently the bonding between the lamina was weak due to the test conditions or to poor fabrication techniques and the failure process degraded the interlaminar bond entirely over a portion of its surface area. Where bonding was good between the lamina, fiber breaks in one lamina tended to induce fiber breaks in the adjacent lamina as seen in figure 13. The explanation for this phenomena is twofold. When a fiber breaks, the loading

condition in the immediately adjacent fibers is increased because the response of the matrix allows load transfer to adjacent fibers. When the interlaminar bonding is effective, the stresses can be transferred across the lamina to fibers in the adjacent layer and fiber degradation in that layer can occur. Additionally, if the interlaminar bonding is adequate, the strain wave initiated by a breaking fiber can be transferred across the lamina and cause the fibers in the next lamina to fail.

Examination of samples from each of the six thermal treatments indicates that, regardless of the severity of the treatment, the matrix material fails by microvoid coalescence. The 6061 aluminum alloys are able to attain a high degree of ductility even after undergoing aging treatments and mechanical working. For 6061 aluminum alloys tested in the as fabricated condition, it has been found that the elongation to failure (percent in 2") in a 1/2" cylindrical sample exceeds 25%. Materials heat treated to the T4 condition exhibit elongation to fracture of 22% and the T6 temper condition has an elongation to failure of 12%. In the composite material, the matrix material is severely constrained by the boron fibers and consequently the effects of thermal treatment including a nitrogen quench which may tend to cause plastic deformation of the matrix are minimized and the matrix

remains very ductile even in the T4N and T6N temper conditions.

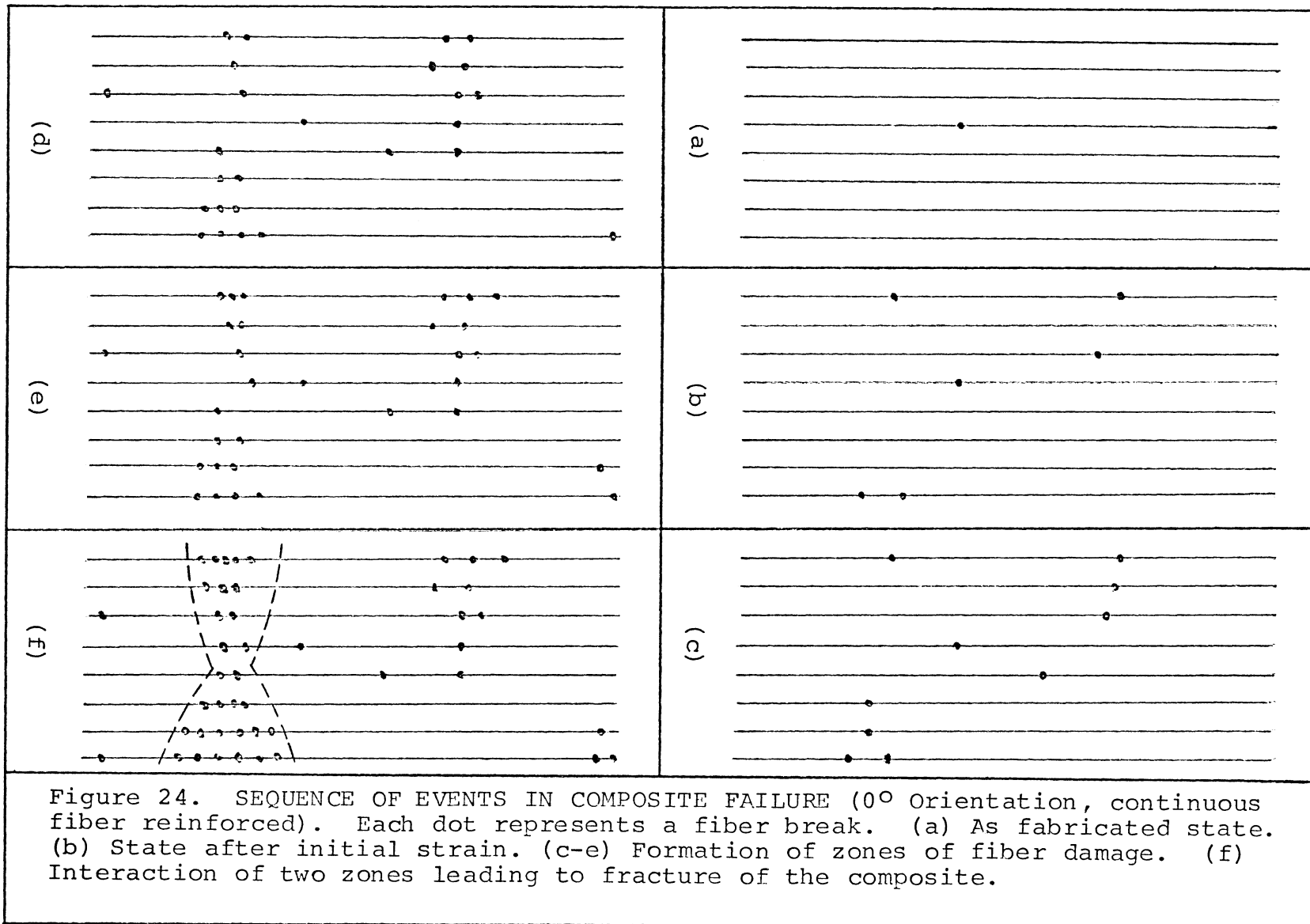
The observations in this study have led to the development of a failure theory which attempts to rationalize the various phenomena observed in the failure analysis. It is believed that the ultimate failure of a composite is preceded by the initiation and propagation of one or more zones of fiber damage. The term damage zone is used to describe an area of multiple fiber damage in which each fiber in the area contains a number of breaks and the distance between breaks is relatively constant.

In the pre-strained condition, a 0° , continuous fiber, boron-aluminum composite has very few fiber breaks due to fabrication or thermal treatment. However, there are a number of irregularities which are inherent to the composite structure and play an important role in the fracture of the material. These irregularities include surface and internal cracks in the fibers, inclusions, node boundaries in the microcrystalline boron, lay-up mistakes in the composite where fibers are not properly aligned, and regions of localized plastic deformation in the matrix. It is also relatively well known that the composite surface constitutes a structural irregularity which often serves as an initiation site for failure.

A composite under load will strain; the fibers and matrix will both strain elastically until a specific deformation is obtained. This specific deformation, ϵ' , is not a unique quantity based solely on structure and properties of the material; it is also influenced by the thermal and mechanical history of the composite and varies considerably from sample to sample. At this specific deformation the matrix will begin to deform plastically. The boron filaments are assumed to be elastic to failure. The plastic strain in the matrix will be localized (as is typical of all crystalline materials). Thus the local shear strains in the matrix will concentrate stresses along the fiber. The fibers begin to break at sites of stress concentration and structural irregularities in the composite. The strain at which the first fiber breaks is designated the critical strain, ϵ^* .

At this stage of development of the failure process, fiber breakage can be considered to be random to the length of the composite assuming that structural irregularities are random. Metallographic evidence indicates that edge fibers and near edge fibers exhibit greater damage than fibers at interior points in the matrix (Figure 3). However, fiber damage does not remain random throughout the entire failure. Once a fiber breaks, the load is transferred both along the broken fiber and from

the broken fiber to nearby fibers by shear stresses in the matrix. The nature of this reaction is dependent on the properties of the matrix and the fiber-matrix interface. When fiber-matrix interactions are minimal, the load can be transferred almost uniformly to all the fibers in the cross sections. Where fiber-matrix interactions are extensive the load transfer can be extremely localized. Thus the mechanical and thermal history of the composite is critical in determining the nature of this reaction. When fiber-matrix interactions are extensive, fibers adjacent to the initial break will begin to support an increased load in an area quite local to the break sites. From this stage until failure occurs, fiber damage due to increased strain is not a random process but a zone process. The nature of the stress concentration in the area of the initial fiber breaks makes the development of a failure zone quite likely. As the strain continues to increase, the composite experiences more fiber breakage, ineffective lengths in fibers are reached in localized areas, and damage zones are created (See Figure 24 for schematic representation). Damage zones grow in size as the strain in the composite is increased and the matrix in a damage zone begins to deform extensively because it cannot support the load. Ultimately as individual damage zones get larger, several zones will overlap



and interact. Metallographic studies of 0° layups indicate that tensile failure is precipitated by the interaction of two or more damage zones. With this zone interaction the composite develops areas perpendicular to the applied stress axis which contains massive fiber damage, and in fact areas which appear to contain only fiber segments of an ineffective length. At this stage the remainder of the fibers fail in tension due to overload conditions. Ductile failure of the matrix occurs and the composite separates into two pieces.

VII. THE FAULT TREE

The working fault tree, as developed for tensile failure of a continuous fiber reinforced composite, is shown in Figure 25. Failure is defined as separation of the composite into at least two distinct sections. Thus the system is defined as all the events and occurrences, whether normal or fault, which can lead to the separation of a composite material. The investigation has involved only 0° continuous fiber reinforced composite materials pulled to fracture in tension. Events such as delamination and interlaminar failure have not been considered. Similarly the mechanical behavior and fracture of composites will be different for fiber and for matrix dominated systems. Fiber domination is assumed by the use of an inhibit gate which restricts use of the branch of the tree to systems in which the product of fiber strength and fiber volume fraction approaches or exceeds the composite strength. The inhibit gate restricts the output of an "or" gate in which three distinct branches characterizing fiber dominated composite behavior have been developed.

The basic events on the left side (branch A) of the "or" gate are: (1) load applied, (2) strain exceeds some critical strain ϵ^* , (3) matrix has low fracture toughness, and (4) inherent structural defects in

composite. In composite systems in which the fiber material is elastic to failure, the critical strain, ϵ^* , cannot be characterized uniquely because strengths of fibers vary over a wide range. This variation is fiber strength, which can be characterized statistically (6,7,8), is the result of inherent defects in the composite material. Once ϵ^* is obtained the first fiber breaks. Due to the brittle nature of the matrix material, the breaking of a fiber induces a matrix crack which is not blunted and thus propagates through the matrix, breaking the fibers in the crack path. It is possible for the matrix crack to propagate through the cross-section of the composite causing the catastrophic failure of the material at a strain of ϵ^* . The failure mechanism described above would lead to a co-planar fracture surface with no fibers extended above or below the fracture surface. No fiber would contain more than one fracture and the failure stress for the composite should be approximately the product of the weak link ultimate strength of the fiber and the fiber volume fraction.

The mid and right branches of the fault tree (branches B and C) leading to the inhibit gate for fiber dominated behavior of the composite involve the transition to a ductile matrix material such as is found in the boron-aluminum or any metal matrix composite system. The basic

events in the right branch (branch B) are: (1) load applied, (2) strain exceeds some critical strain ϵ^* , (3) matrix has high fracture toughness, and (4) inherent structural defects in composite. The critical strain will depend on the fiber properties and when considerations are restricted to only those cases where the matrix does not greatly affect fiber behavior, ϵ^* will be the fracture strain of the fiber. When ϵ^* is obtained, the first fiber breaks. Since fiber matrix interactions are minimal the first fiber break occurs in the fiber which has the lowest strain to failure. Immediately, the load on all other fibers is increased and the strain in the composite is increased. Because fiber-matrix interactions are minimal, this increase in the loading condition is uniformly shared by all the fibers in the cross section. The increase in strain may be sufficient to cause the second fiber to break or an additional strain perhaps due to an increase in the applied load may be required before second fiber breakage occurs. It is important to emphasize that the matrix material has sufficient fracture toughness to blunt a matrix crack. Second fiber breakage is followed sequentially by a third, fourth, etc., fiber break until all fibers have failed and the composite separates. Each fiber break would occur randomly relative to the preceding break, the break site being controlled

by the strength and flaw characteristics of the individual fibers. Such failure would lead to an extremely rough fracture surface with fibers of various lengths above and below the fracture midplane. As in the case of a composite with a brittle matrix material no fiber should contain more than one fracture. The failure stress for the composite should be approximately that predicted by the rule of mixtures: the product of the ultimate strength of the fiber times the fiber volume fraction plus the product of the ultimate strength of the matrix times the matrix volume fraction.

The mid branch of the fault tree (branch C) involving fiber dominated behavior describes the condition in which there is extensive interaction between the fibers and matrix. The basic events in the mid branch (branch C) of the fault tree are (1) load transfer capabilities good between fibers and matrix, (2) final thermal treatment below test temperature, (3) $\alpha_m \gg \alpha_f$, (4) strain exceeds some critical strain ϵ^* , (5) load applied, and (6) inherent structural defects in composite. This branch describes the events which occur due to exposure of the composite to various thermal and mechanical conditions leading to extensive fiber-matrix interactions. All metallographic analysis and a search of the literature indicates that thermal treatments of the composite inducing either

chemical or mechanical reactions tend to alter the bonding characteristics between the fibers and matrix and that this change in bond character changes the fiber response when the material is tested. In no case has it been found that where adequate bonding existed between the fibers and matrix prior to thermal treatment did an exposure to treatment improve the strength of the composite. The event "fracture of fibers at a higher stress than predicted without fiber-matrix interactions" appears in the fault tree because there are no obvious reasons why fiber-matrix interactions, whether chemical or mechanical, should only lower the apparent fiber strength. However, an evaluation of literature data indicates that one general aspect of fiber-matrix interactions is a lowering of the apparent fiber strength. Consequently the increase event has been left in the fault tree with the stipulation that it applies to no real composite system.

The significant event leading to separation in this branch is "fiber fracture occurs at lower stress than predicted without fiber matrix interactions." The apparent lowering results in the failure stress of the composite being typically less than the product of the fiber strength and fiber volume fraction. Lowering of the apparent strength of the fibers can occur because of either chemical or mechanical interactions between the fibers and the matrix.

Mechanical interactions are due to the differences in coefficient of thermal expansion between fibers and matrix ($\alpha_m \neq \alpha_f$) and the thermal treatment. Mechanical interactions to lower the fiber strength would result from either a residual bulk tensile stress in the fiber or from localized interactions between the ductile aluminum matrix and the brittle boron fiber. Only in temper treatments where the final thermal treatment occurs at temperatures below test temperature does the residual stress in the fiber become tensile. This event can be caused by a nitrogen quench or other suitable treatment if the load transfer capabilities between the fibers and matrix are good and if $\alpha_m \gg \alpha_f$. It is necessary that the load transfer capabilities between fibers and matrix must be good or the fibers cannot attain a residual stress state on thermal cycling because stresses cannot be generated between components that are not well bonded. Mechanical interactions, whether the results are localized in the matrix or retained in the stress state of the fiber, depend on intimate contact at the interface due to optimum fabrication conditions. Due to the irregular nodular surface of the boron fibers in the boron-aluminum composite system there is gross seizure between the fibers and matrix in that system. The effects of quenching which causes the matrix to be loaded in longitudinal and radial tension

relative to the fibers greatly increases the effects of seizure between the fibers and matrix material. Early failure of a brittle fiber can be induced by the combined effects of tensile loading and stress concentrations in the matrix at various flaws and defects along the fiber length. Stress concentrations along the fiber length are the result of localized elastic or plastic strains in the matrix. Strain in the matrix is caused by thermal treatment and differences in the coefficients of thermal expansion ($\alpha_m \neq \alpha_f$). When concentrated at defect sites in the fiber, residual strains tend to reduce the strength of the fiber and initiate early fracture. Defect sites of importance are composite surface, inclusions and node boundaries (in boron) and surface flaws in fibers, surface and internal cracks in fibers, and lay up mistakes in the composite.

Chemical interactions between components can be produced by the diffusion controlled solubility of the fiber and matrix material. Additionally chemical reactions can be induced by the formation of a reaction product at the interface. Regardless of the mechanism of chemical interaction or the products found, the result is a lowering of the strength of the fiber and failure initiates at lower strains.

When the composite is treated by a high temperature excursion and cooled to test temperature, the fibers are loaded in residual compression and the matrix in residual tension in the longitudinal and radial directions. Although the composite still fails at a lower stress than predicted without fiber matrix interaction, it is believed that the stress state in the fiber is of no consequence. In this case early failure of fibers occurs due to fiber-matrix interactions and by the thermal treatment. These interactions involve diffusion of soluble species, formation of a reaction product at the interface and elastic and plastic strains in the matrix adjacent to the fibers. Each of these fiber-matrix interactions tends to lower the strength of the fiber and induces early failure of the composite relative to a material whose thermal history is not so extensive.

An additional sequence of events that would cause the fibers to be loaded in residual tension would involve a situation where (1) the coefficient of linear thermal expansion in the fiber is greater than that in the matrix ($\alpha_f > \alpha_m$) and (2) the temperature of the final thermal treatment is above test temperature. A residual tensile stress in the fibers would occur assuming that the load transfer capabilities were good between the fibers and matrix. This occurrence of events would result in

fiber-matrix interactions to reduce the load carrying capacity of the fibers. Although this sequence may be of some theoretical importance, there is no known composite system where α_f is greater than α_m . Consequently the sequence appears in the fault tree with the stipulation that it applies to no real composite system.

Development of the mid branch (branch C) of the fault tree from the event "Fracture of fibers at lower stress than predicted without fiber-matrix interactions" to "separation" involves the initiation and propagation of damage zones in the composite material as described in the discussion section of this thesis.

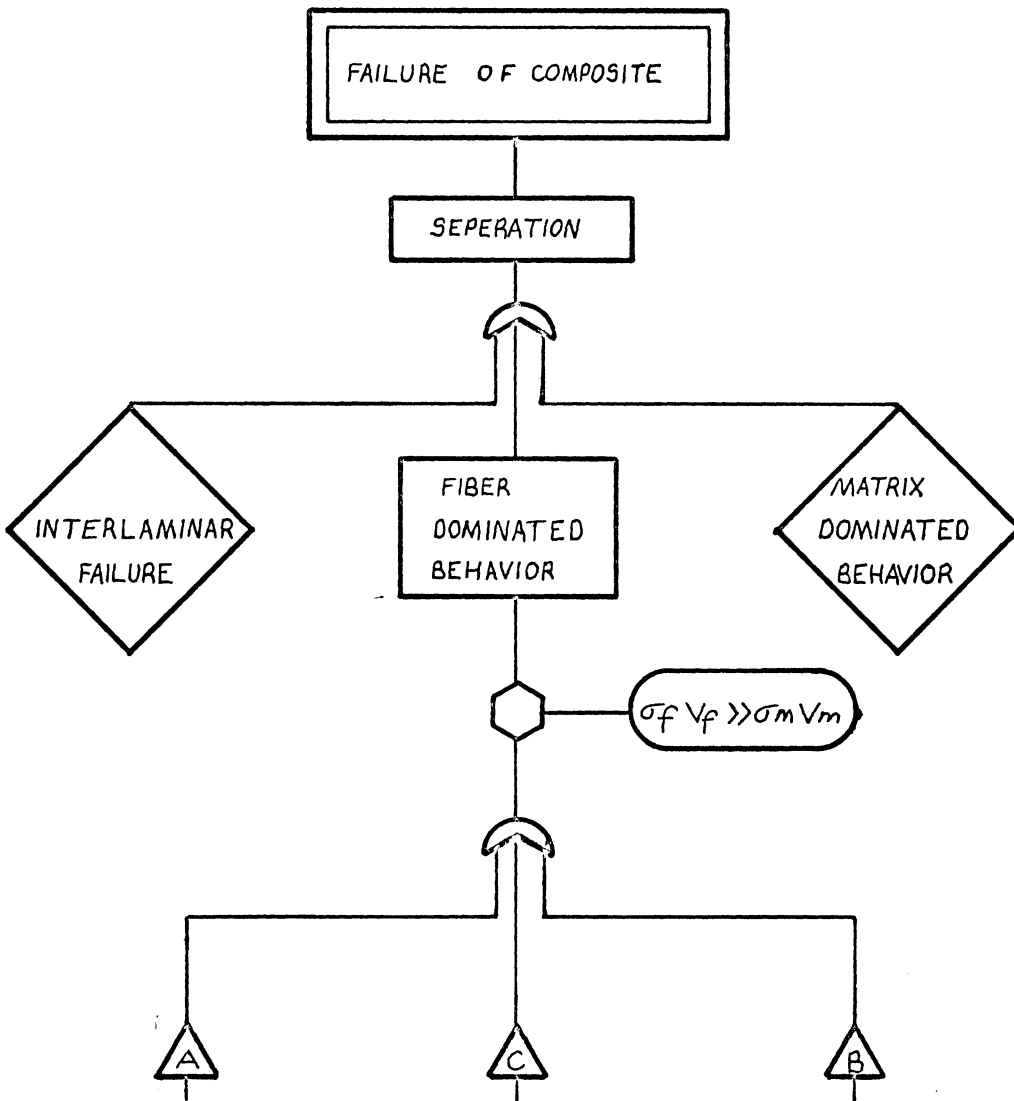


FIGURE 25. THE FAULT TREE

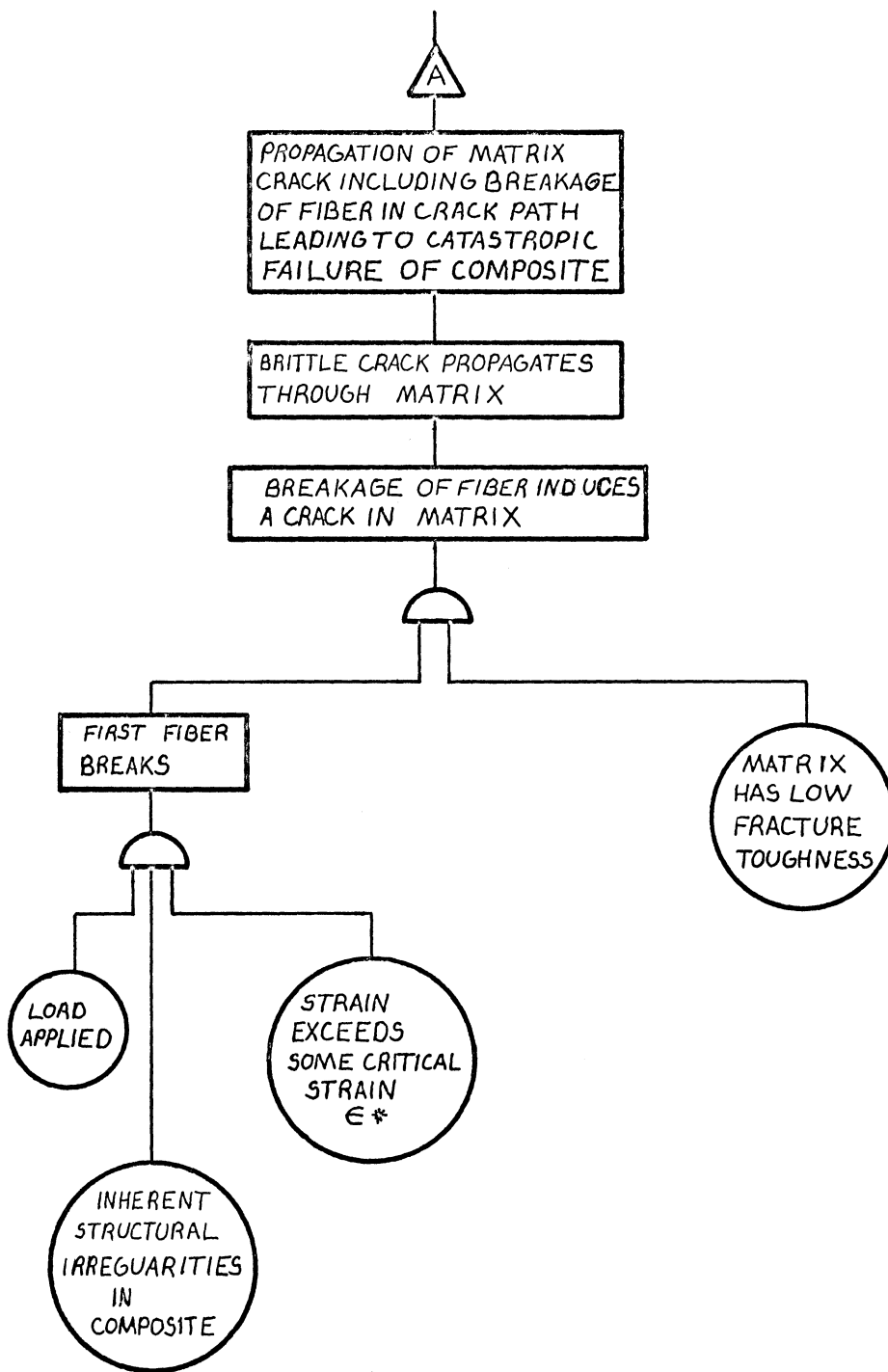


FIGURE 25 (CONTINUED)

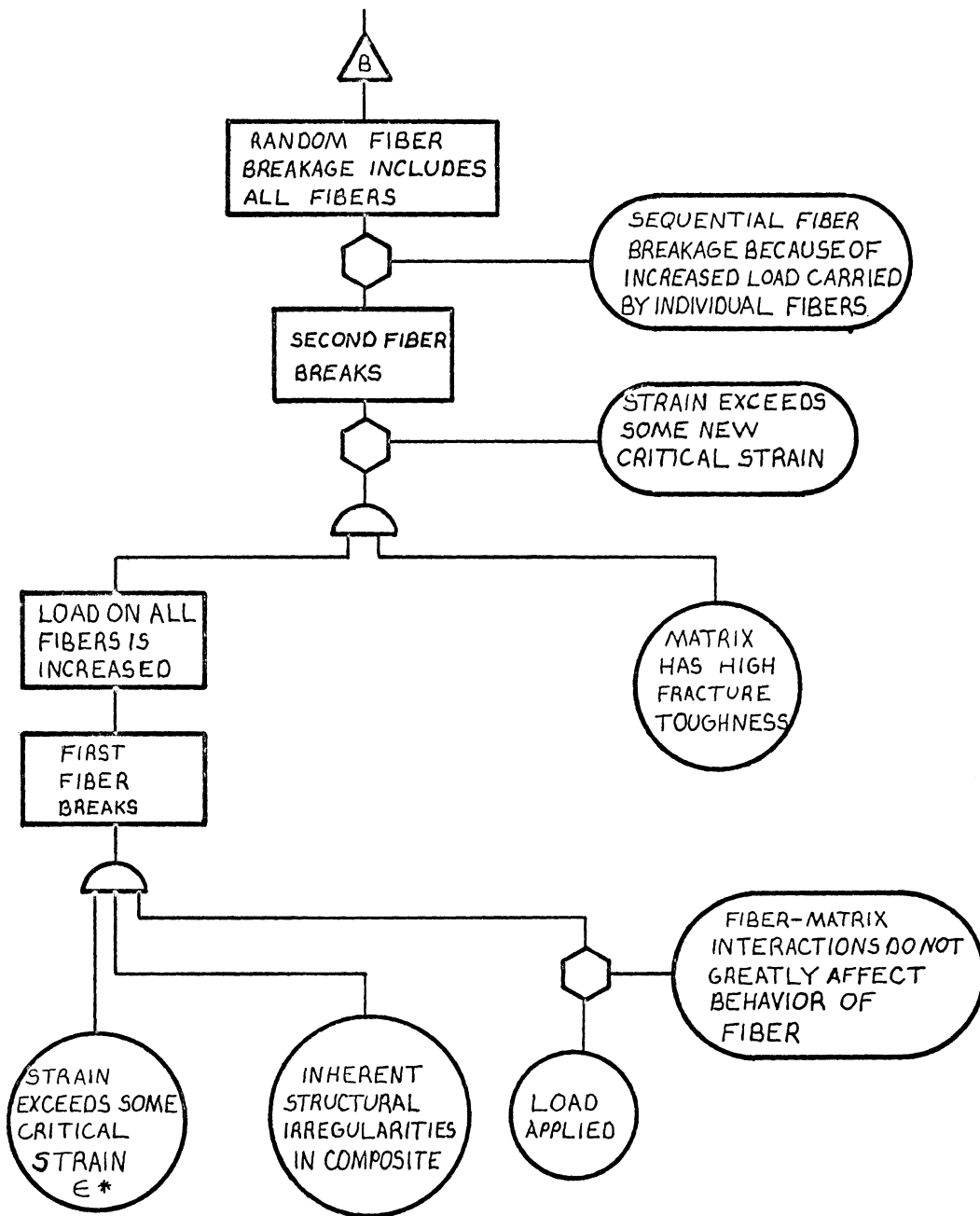


FIGURE 25 (CONTINUED)

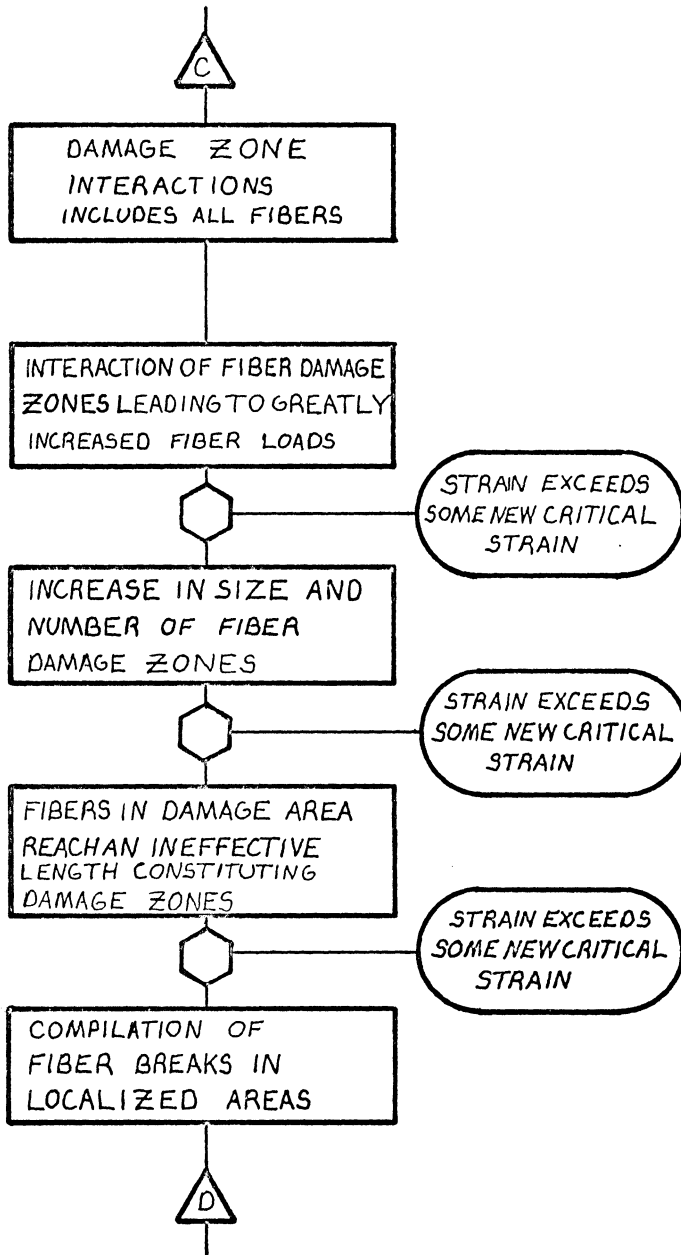


FIGURE 25 (CONTINUED)

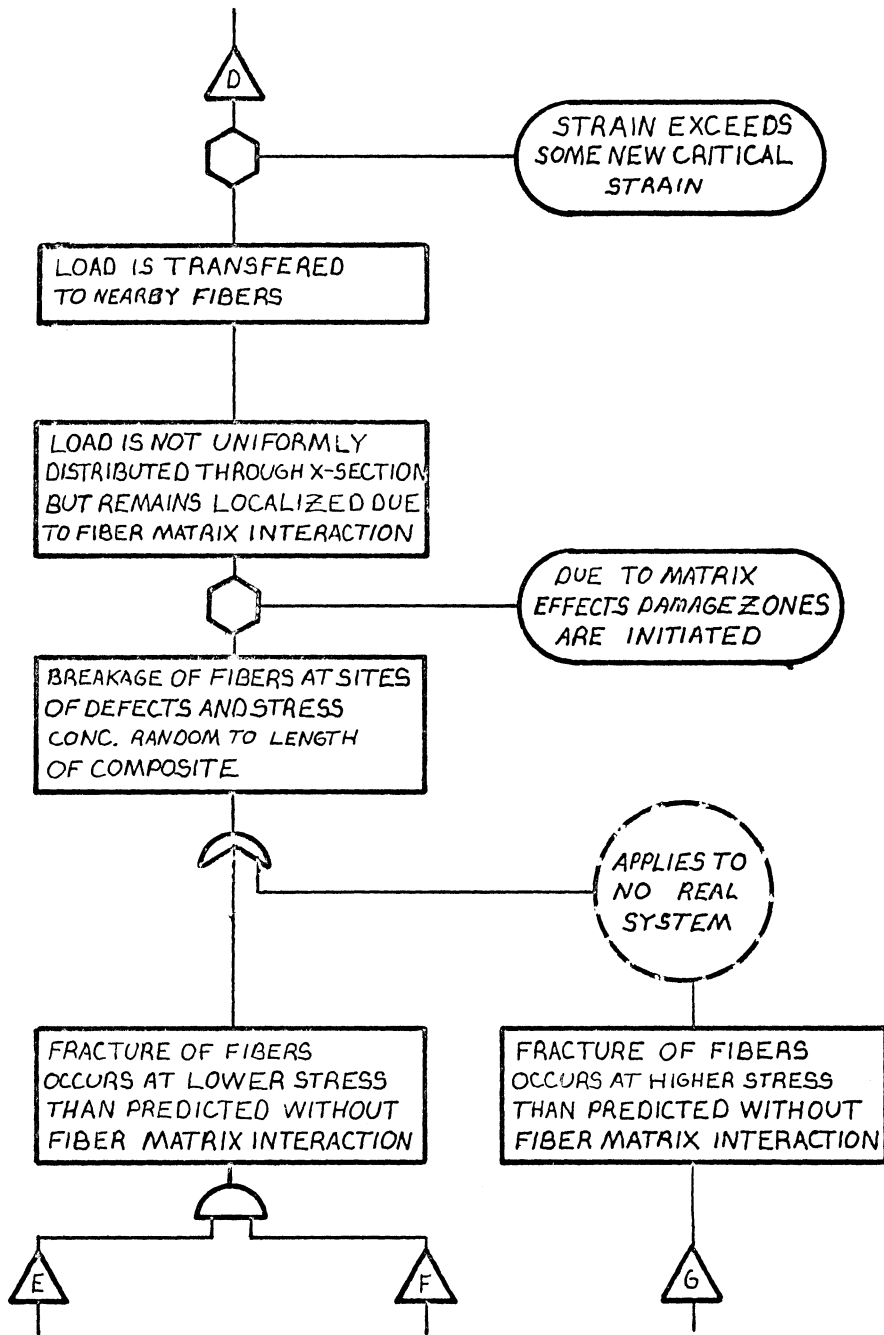


FIGURE 25 (CONTINUED)

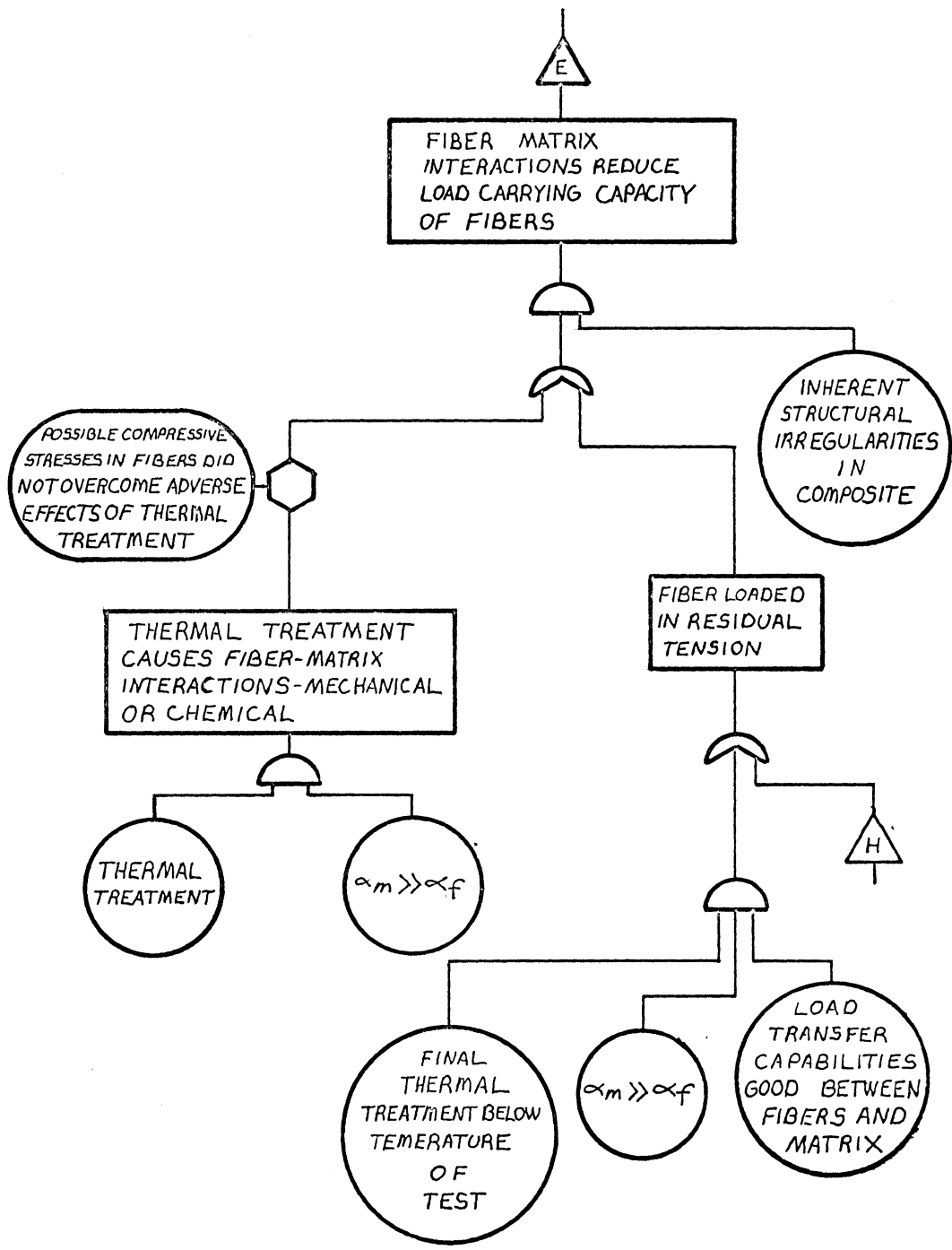


FIGURE 25 (CONTINUED)

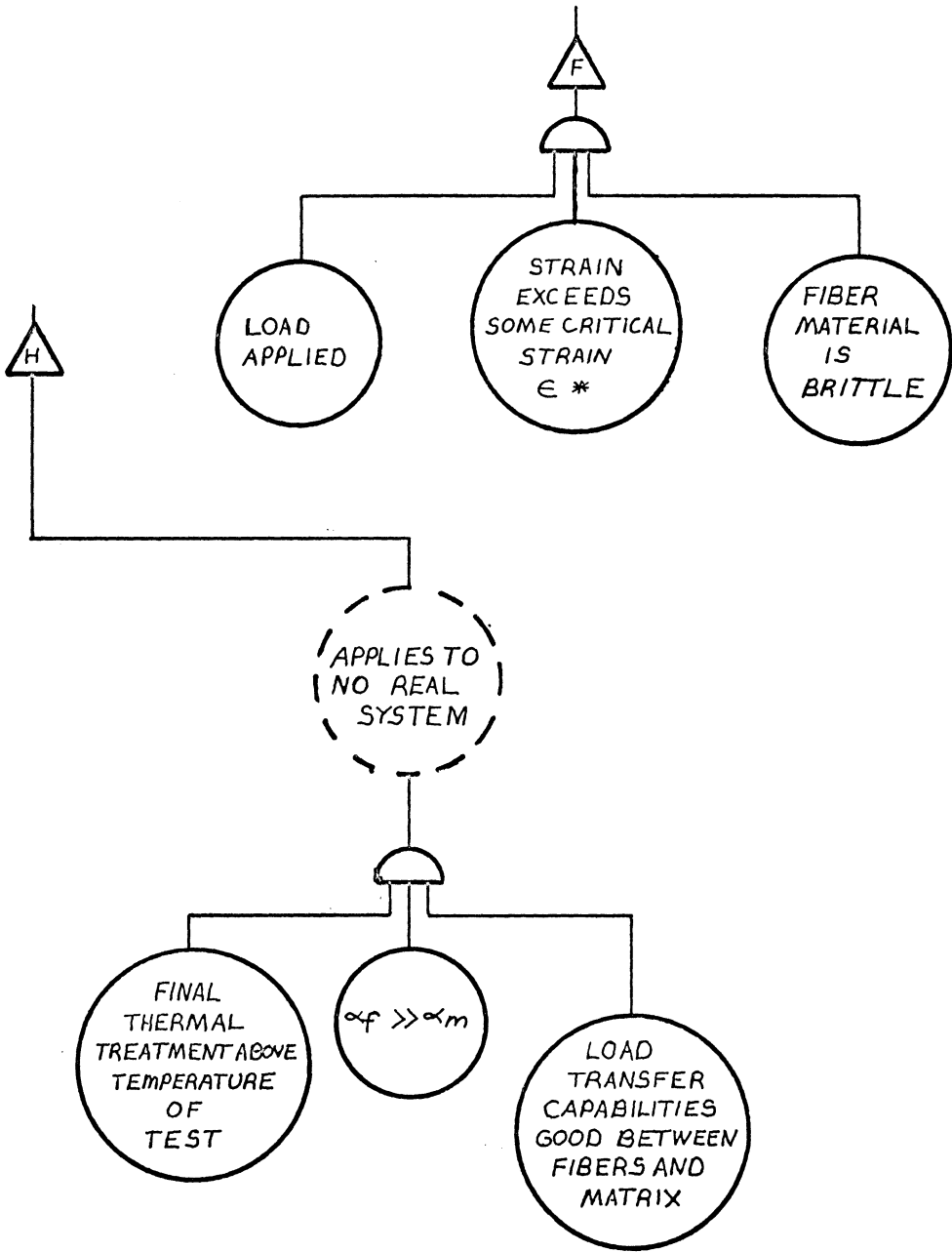


FIGURE 25 (CONTINUED)

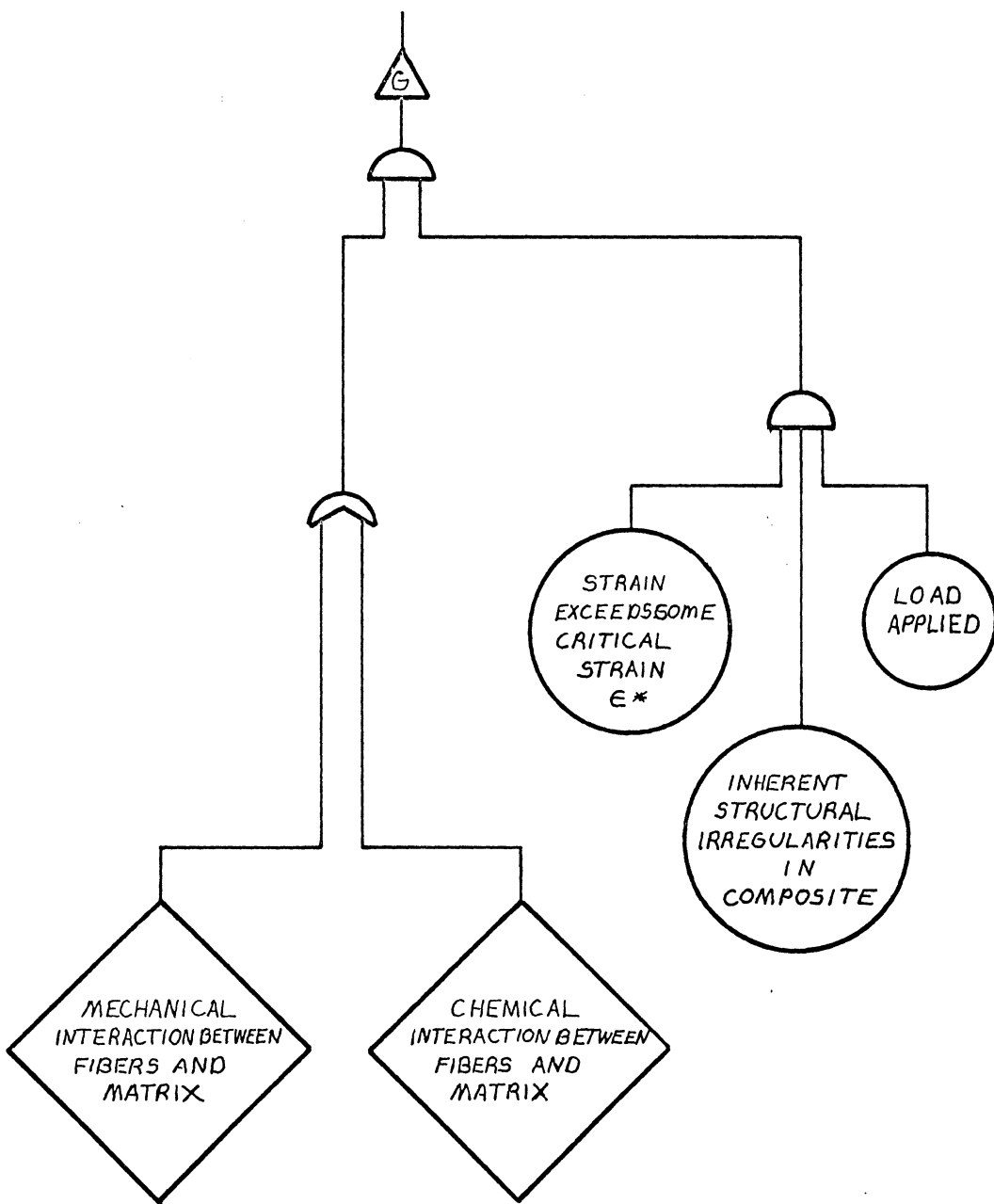


FIGURE 25 (CONTINUED)

VIII. CONCLUDING REMARKS

The results of this investigation have shown that the tensile failure of unidirectional composite material can occur by one of several mechanisms depending on the properties of the material and the previous thermal and mechanical conditions to which the material has been subjected. When the matrix material is brittle and can be characterized by a low fracture toughness, the failure of a single fiber can be sufficient to initiate and propagate a matrix crack through the material leading to the catastrophic failure of the composite at a low stress. This low stress to failure can be considered to be the product of the weak link ultimate stress of the fiber material and the fiber volume fraction. In composite materials where the fracture toughness of the matrix is high, it is likely that matrix cracking will not be an effective event in the failure of the material. In this case the strength of the material is determined by the degree of interaction between the fibers and matrix. In composite systems where the fiber-matrix interactions are minimal, the composite failure strength approaches that predicted by the rule of mixtures: the product of the ultimate strength of the fiber times the fiber volume fraction plus the product of the ultimate strength of the matrix times the matrix volume

fraction. When fiber-matrix interactions become extensive, as is the case in the boron-aluminum system, the strength of the composite material is significantly reduced.

Fiber-matrix interactions can be chemical and/or mechanical in nature and the effect of either is to lower the failure strength of all composite systems where fiber dominated properties exist. In composite systems where fiber-matrix interactions are extensive, failure occurs by the initiation and propagation of fiber damage zones. Due to extensive interactions between the fiber and matrix, the load is not uniformly shared by all fibers in a given cross section and the breakage of a single fiber tends to concentrate the load locally rather than uniformly in the cross section. Fiber breakage is not random but zone oriented. Fracture of the composite occurs by coalescence of zones of fiber damage at strengths typically less than predicted by the rule of mixtures.

It has been demonstrated in this thesis that the fault tree methodology is a valuable analytical tool in the evaluation of the failure mechanism in each case. This investigation has identified those chains of events leading from basic events to the failure event and has defined areas requiring further investigation. The observations included in this thesis indicate the general methodology required for understanding the fracture process and provide

a useful technique for failure analysis. The graphical description of the system, the events in the system, whether normal or fault, and the basic events which can lead to failure are all useful as an aid to understanding the system and as a tool for determining the paths that lead to failure. The application of the fault tree methodology to failure process and analysis can provide significant insight into the process and can prove to be a general methodology for failure analysis.

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FAULT TREE ANALYSIS OF FAILURE IN
BORON-ALUMINUM COMPOSITES

by

Thomas Smith Allen

(ABSTRACT)

Boron reinforced aluminum composite materials are useful in many aerospace applications due to their improved physical properties such as high specific strength and reasonable thermal stability at elevated temperatures. The shear strength, high transverse stiffness and fracture toughness of the matrix are also important material properties. Failure mechanisms in the boron-aluminum composite system are often difficult to characterize because not all failures initiate in or remain limited to the area of final material fracture. Furthermore, because of the extremely brittle nature of boron fibers, ordinary sample preparation cannot be used for metallographic analysis. Frequently, the use of abrasive techniques, such as grinding and polishing, further degrades the material, making an accurate assessment of fiber damage difficult at best and possibly even misleading. To overcome this difficulty, fault tree methodology was applied. Fault tree methodology was combined with chemical leaching techniques to determine

the types and degrees of fiber degradation in boron-aluminum composites. Samples with various thermal histories were examined, using optical and electron microscopy, and a fault tree was constructed which qualitatively characterizes failure development in these composite materials.