TECHNICAL INFORMATION SERIES

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SOME FUNDAMENTAL FRACTURE MECHANISMS APPLICABLE TO ADVANCED FILAMENT REINFORCED COMPOSITES

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SPACE SCIENCES LABORATORY

MISSILE AND SPACE DIVISION

GENERAL ELECTRIC

CONTENTS

PAGE

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LIST OF FIGURES

ABSTRACT

The object of this study is to establish experimentally the critical fracture modes of an epoxy matrix in the vicinity of a break in a high modulus, high strength filament. Three distinct failure modes were observed to occur and the nature of these three modes is explained through an analysis of the stress state in the matrix. A single filament embedded in an epoxy novolac was used to examine the fracture process. The advantage of using this type of resin arises from its ability to be modified to varying degrees of crack sensitivity.

The effects of varying gross strain rate were evaluated and the study was then extended to specimen configurations containing more than one filament. This approach permits the examination of the interaction of nearby filaments with localized fracture processes. Filaments of boron on tungsten (B/W) boron on Silica (B/SiO_2), boron carbide on B/W and tungsten wire were used with epoxy novolac (DEN 438).

I. INTRODUCTION

Recent technological advances in structural materials have indicated the potential of filament reinforced composites for providing greater structural efficiency than high strength metals. Most of the studies reported in the literature have attempted to establish the strength characteristics of these materials under static loading conditions with relatively little emphasis on the mechanics of the failure process. Since composites derive their strength from the reinforcement of a low modulus matrix with high-modulus filaments, a major objective in fabrication has been to increase the volume fraction of filaments to obtain as much reinforcement as possible in a given volume of composite. Both metal matrices (such as aluminum and nickel) and polymer matrices (such as epoxies and phenolics) have been reinforced with continuous filaments of boron-carbide, boron, silicon-carbide, tungsten and carbon.

Some efforts have been made to determine the critical aspects of the reinforcing mechanism by studying simple models both experimentally and theoretically. Dow(1) and Cox(2) performed some of the early theoretical work in this area, and Tyson and Davies(3) and Shuster and Scala(4) conducted photoelastic studies of the stress patterns around the reinforcing filaments. These studies show that the filament is loaded primarily through shear at the interface with some direct tension at the ends depending on end geometry. The magnitude of the shear stress in the matrix varies along the filament and radial to it, with maximum shear stresses occurring at the interface near to the filament ends. Schuster and Scala(4, 5) also studied interaction of the stress patterns at whisker ends and although no quantitative evaluation was made, for actual breaks the isochromatic patterns indicated a higher stress concentration in the matrix near a broken whisker than at a whisker end. It has therefore been established that the formation of ends when a filament fractures is the most critical stress raiser in the matrix material.

Rosen(6) has investigated the behavior of single layer, unidirectionally oriented glass filaments in an epoxy matrix. In this study, the failure of individual filaments was noted during the loading process until total failure of the specimen occurred. Rosen developed a theory of failure relating to cumulative damage and this work has been extended by Friedman[7]. A basic assumption in this theory is that the matrix makes no significant contribution to the tensile strength of the composite but simply provides a means for transferring load to the filaments through shear stress. This introduces the concept of bundle strength. When a bundle of filaments has no matrix to transfer load, the failure of a single filament renders that filament ineffective in supporting any part of the load. Therefore, as the individual filaments reach their tensile strengths and fail, the unbroken filaments must absorb a greater portion of the load until the entire bundle fails. The load which the bundle can then sustain is a function of the statistical distribution of filament strengths in the bundle, if we presume that there is no matrix present. The bundle theory would appear to be conservative when a matrix is used, because the matrix permits a broken filament to reinforce over all of its length except the region of failure. However, should the matrix serve to concentrate the stresses at a filament fracture, it can contribute to the fracture of adjacent filaments.

A great number of studies have been made to establish the strengh characteristics of composites, with success being measured in many instances by the degree to which "rule of mixtures" predictions have been attained. Although the name is somewhat deceiving, this theory is based on the simple assumption of uniform strain in both fiber and matrix and linear elastic response. The form of the relation for modulus is given as

$$E_c = V_m E_m + V_f E_f \quad \text{.....................} \quad (1)$$

where E_c, E_m, E_f are the moduli of the composite, the matrix and the filament respectively, V_m and V_f are the volume fractions of matrix and filament. An analagous relation for uniaxial tensile strength is obtained by replacing the moduli with the tensile strengths of the matrix and filaments to predict the tensile strength of the composite. Several investigators such as Kelley & Davies[8] and McDanels et al[9] report good agreement with "rule of mixtures" predictions for modulus and tensile strength when metallic filaments are used. However, most studies report somewhat less agreement for predicted strengths, especially in the region of high filament content, and the agreement is even less apparent in systems using continuous and discontinuous non-metallic filament arrays.[10][11][12] Deviations from rule of mixtures predictions are often attributed to inadequate fabrication techniques resulting in voids, fiber-to-fiber contact, poor bonding, fiber degradation and the like.

Hashin[13] used limit analysis methods to bound the yield surface of multiphase materials which obey Mises yield conditions. The lower bound on the yield surface of the composite is simply the strength of the weakest phase. However, the upper bound for simple tensile strength of the composite is given by

$$K^* = \sum_{r=1}^{r=n} K_r V_r \qquad (2)$$

where:

K^* and K_r are the strengths of the composite and the r constituents, respectively

V_r are the volume fractions of the r constituents.

This upper bound on strength is essentially the rule of mixtures relation and is of particular interest in the light of Hill's[14] work relating to modulus. The latter shows that the modulus of an idealized composite has a <u>lower</u> bound defined by the rule of mixtures, depending on the differences in Poisson's ratio of the constituents. This suggests that fundamental variations in constituent properties are reason enough for considerable deviation from rule of mixtures behavior. The need for a better understanding of gross composite behavior is clear, but to establish strength characteristics the mode of failure is a primary concern. This study deals with the essential characteristics of filament and matrix failure using advanced filaments. Because of their extensive use in advanced composite structures, boron filaments exhibiting tensile strengths of 300 to 400 thousand psi were used. The resin system was epoxy novolac. Specifically, the fracture processes in filament and matrix were examined with the goal of establishing a basis for improving resistance to crack propagation in composite materials.

To minimize ambiguity in the observation, the specimens used in this investigation were simple: two-inch long filaments were imbedded in epoxy specimens having a gage length slightly in excess of two inches; the cross sections were approximately 0.080" to 0.250" wide by 0.070" to 0.080" thick. Single and (for long range interaction effects) multiple filament specimens were loaded in tension in an Instron machine operated over a range of strain rates ($\sim$.01 to 1 in/in/minute). Owing to the transparency of the epoxy matrix, filament cracking could be visually observed and (at low strain rate and high chart speeds), correlated with transient variations in the load-elongation curves. Being thin, the specimens were suitable for post-test examination under the microscope using vertical, oblique and transmitted illumination.

II. FUNDAMENTAL FAILURE MECHANISMS

In the past, two modes of local failure have been observed in the fracture surfaces of composites: fiber fracture and fiber pull-out. The first mode is essentially fiber limited while the second suggests that insufficient bonding was available to fracture the fiber. Unfortunately, in a densely packed composite the degree to which the two modes are discernible depends on how tortuous the failure process is.

This points up the desirability of observing the local failure process in very simple specimens. This is precisely the approach here and, using this simple technique, three modes of local failure were observed in the matrix surrounding a filament fracture.

To understand the conditions which contribute to the fracture process, it is first necessary to establish the stress state in the matrix while the composite is loaded in tension. For this analysis it is useful to consider the stress variations described in Figure 1.

In addition to the variation of shear stress along the interface, there is a stress gradient normal to the filament axis which has been determined from the photoelastic investigations referenced above. This variation is best described by plotting the tensile stress trajectories in the matrix as determined by Tyson and Davies[3]. In Figure 2, the direction of the tensile stresses is tangent to the trajectories at a given location and the magnitude of the tensile stress is proportional to the proximity of the trajectories. The figure shows that the greater concentration of stress transfer from matrix to filament occurs near the end, while at the center of the filament (element A), the matrix is virtually unstressed, since the stiffer filament carries most of the load. At element B (slightly removed from the geometric discontinuity of the end), the stress state is nearly pure shear and the tensile stress trajectory makes an angle of approximately 45° with the filament.

At element C there is considerable tensile stress in the matrix parallel to the filament axis in addition to shear stress at the interface. The inclination of the principal stress trajectory to the filament is therefore small at the ends.

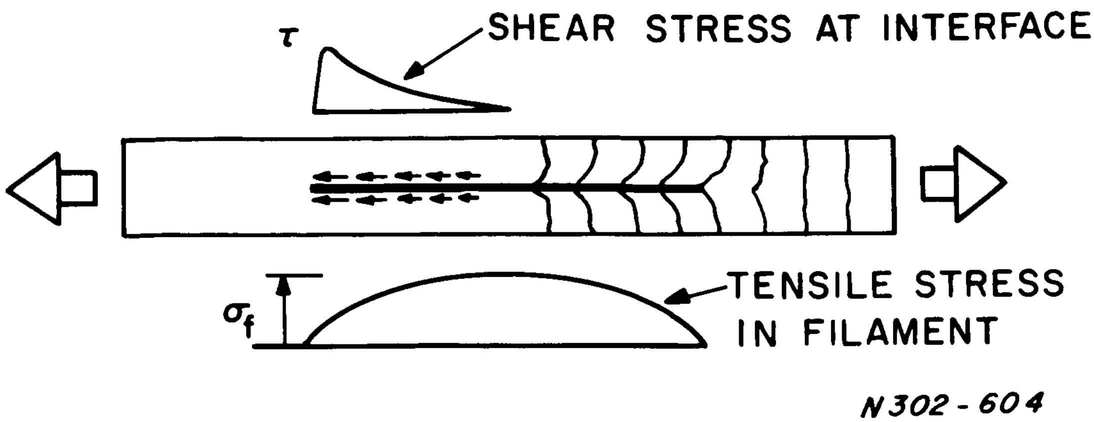


Figure 1. Variation in Interface Shear and Filament Tensile Stress

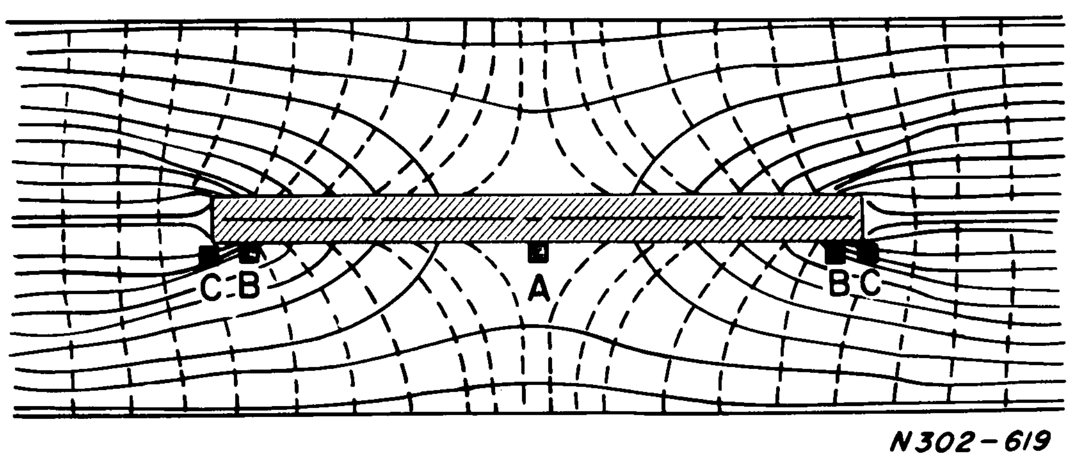


Figure 2. Tensile Stress Trajectories in the Matrix (Reference 2)

Shuster and Scala[4] have analyzed the shear stress distribution at the end of the filament where the geometric discontinuity exists. The interface shear stress diminishes somewhat very near to the end point and therefore is not as critical as at element B which is slightly removed from the end of the filament.

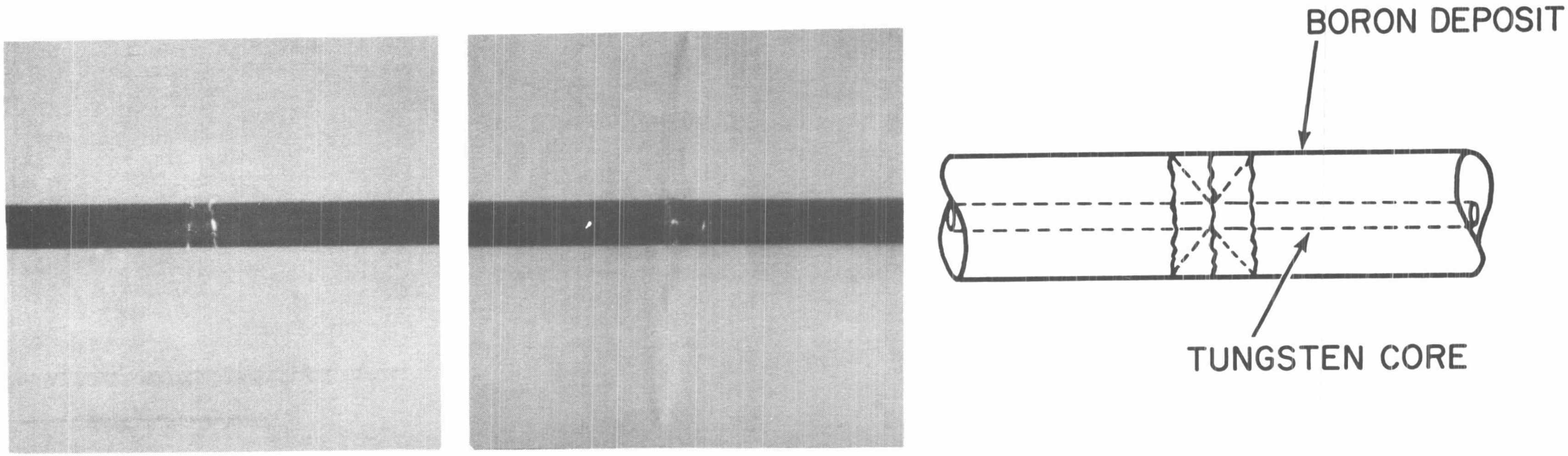
With the stress state in the matrix qualitatively described, the possible modes of failure can now be defined. The filament is highly stressed along most of its length to within 5 diameters of the ends. It is in this region that the first fracture occurs, the specific location being dependent on the variation in strength along the filament. The matrix adjacent to the filament at the fracture is suddenly stressed (having been virtually unstressed) to carry the tension previously carried by the filament. At the exact location of the filament break, this stress state is primarily tension in the direction of the filament axis and as the tensile strength of the matrix is exceeded, a disk-shaped crack is formed at that point normal to the filament axis. Presuming sufficient bonding between matrix and filament, the extent which the disk-shaped crack propagates depends on the ability of the matrix to absorb energy at the increasing circumference of the cracked region. In some resins, this initial filament fracture results in failure of the entire specimen, but it is possible to formulate a resin which can absorb the energy sufficiently to arrest the crack caused by the filament fracture.

In some instances, the filament fractures at low load at a point of weakness and insufficient energy is released to produce a matrix crack. This condition may be termed a low energy fracture and is illustrated in Fig. 3A, while a somewhat higher energy fracture results in a disk-shaped crack as illustrated in Fig. 3B. The terms "low-energy" and "high-energy" failure are used here to make an inferred distinction between manifestly different combinations of filament matrix behavior. When a brittle filament fractures, the elastic strain energy, which is in excess of that required to create the fracture surface (in the filament), is instantaneously released and must be absorbed or transmitted by the matrix in the vicinity of the filament failure. For a given filament, the magnitude of this energy pulse will be approximately proportional to the filament tensile strength; and it will vary from point-to-point in a manner consistent with the strength-statistics of the filament. Thus, one element of the "low-energy" or "high-energy" designation is a function of the filament strength.

The matrix response is the other element in the designation; that is, it must respond instantaneously to a rapidly propagating internal crack having the dimensions of the filament cross section and characterized by a high stress and strain rate at the crack tip. In the (assumed) absence of plastic or visco-elastic deformation at this local strain rate, the energy pulse may be partially absorbed by damping during its transmission (as an elastic wave) to the specimen mass. On the other hand, the energy pulse may be partially absorbed by the creation of new surface area as the filament crack propagates into the matrix. The "crack sensitivity" of the matrix will determine the magnitude of the energy pulse required to propagate such a crack. The formulation and cure cycle of the matrix used in this work had a "critical" crack sensitivity in that it maintained or lost its local integrity in a manner consistent with expected local variations in filament tensile strength.

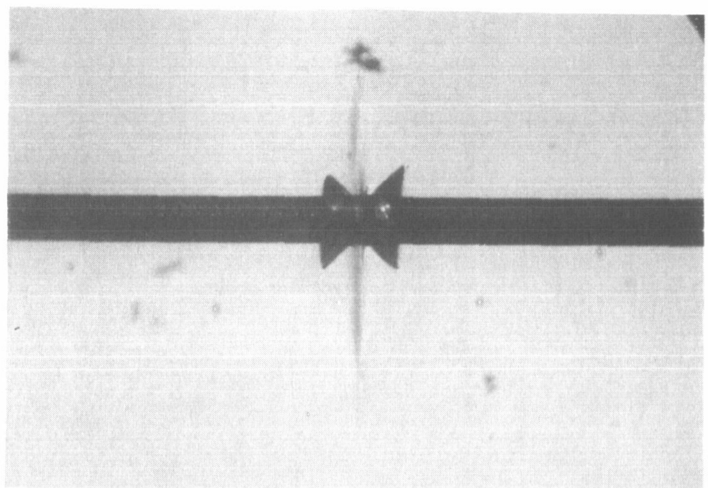
The most common form of fracture in a boron filament shown in Figure 3C is comprised of two cup-cone fractures emanating from the filament substrate and a single normal plane of fracture in the center section. This fracture pattern can result in more than one crack emanating from the filament into the matrix at the time of initial filament fracture as shown in Figure 4a.* Once the sudden crack initiation process has occurred, the the propagation is determined by the new state of stress in the region of the crack tips. This stress state is very nearly that described previously for the end of a filament except that there are two such ends back to back and the end geometry is somewhat more complicated. Since there is no direct tension on the newly formed ends, a pure shear condition exists at the

* The photographs presented here are for the unloaded condition. The geometry changes somewhat under load and this will be illustrated later in the discussion.

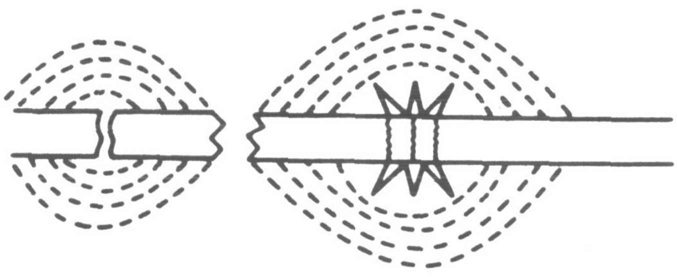


(a) (b) (c)

Figure 3. Fracture of a Boron Filament in Epoxy Novalac
(a) Low Energy Fracture (58X) (b) High Energy Fracture (58X)
(c) Schematic View of Fracture Pattern Detail



(a)

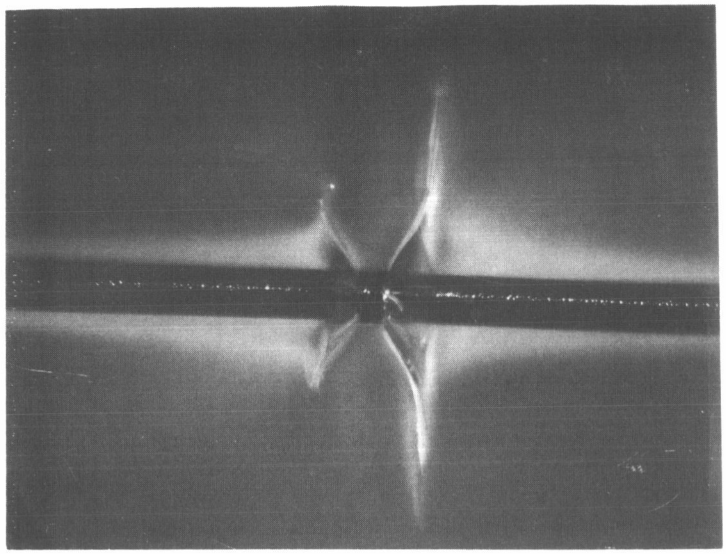


(b)

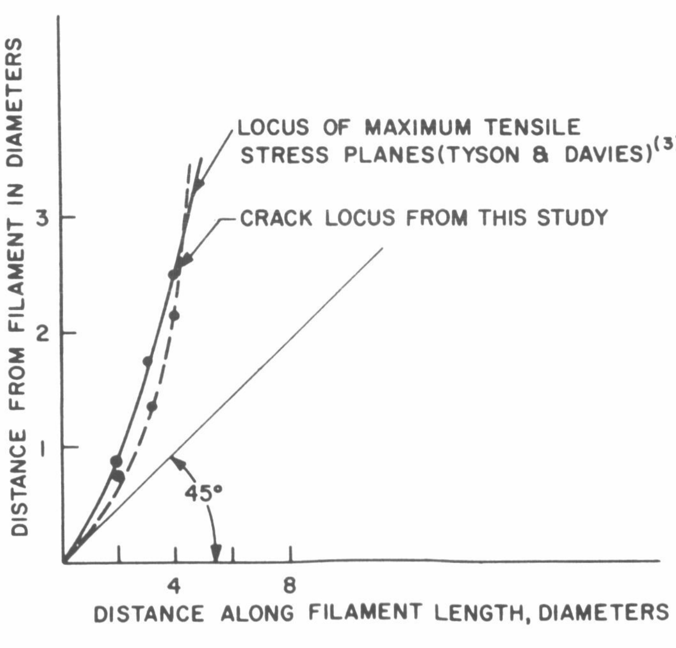
Figure 4. Multiple Crack Site at Filament Fracture in Epoxy Novolac (a) Photograph at 58X (b) Schematic Diagram of Most Probable Stress Trajectories which occur for single and multiple crack sites.

interface a short distance from the fracture location. The disk-shaped crack separates the two end stress patterns and represents a free surface across which no tensile stress can be transmitted. Tensile stress trajectories are again a useful tool for describing the change in stress transfer when a fracture has taken place. In the case of a low energy break where no significant initial cracks are formed, the stress trajectories at the break might appear as shown in Figure 4b, extending by as much as 10 to 15 filament diameters from the break in each direction. When a high energy fracture exists and a disk-shaped crack is formed, the trajectories which would normally pass through the crack region are no longer effective and the pattern must change. This essentially results in a pattern of stress concentration at the crack tip and a region immediately adjacent to the crack source which is virtually unstressed. This is the region in which the inclined cracks have been formed and hence the inclined cracks are to some extent shaded by a larger disk crack, preventing them from propagating as quickly as the disk crack. When the disk-shaped crack does not extend further into the matrix than the inclined cracks (which form truncated cones), then there is a stress field at the tip of the inclined cracks and they begin to propagate in that field.

The direction of propagation of the disk-shaped crack in the absence of other filaments is more or less normal to the filament axis. This is the locus of planes of maximum tensile stress in the matrix of the filament fracture and is what one might expect, given the tensile stress trajectories. When the disk shaped crack does not shade them, the inclined cracks also propagate in a direction normal to the tensile stress trajectories. This results in a curved crack becoming nearly normal to the filament within a radial distance of 3 or 4 diameters from the interface. Figure 5a illustrates this at a location of low energy fracture (note the absence of a disk). Figure 5b shows a plot of the crack locus superimposed on the typical stress trajectories constructed from the data of Tyson and Davies[(3)]. In the photo of Figure 5, the two inclined cracks begin propagating simultaneously, but at some point in the propagation process, one crack overshoots sufficiently to



(a)



(b)

Figure 5. Propagation Pattern for Inclined Crack in Epoxy Novalac (a) Photograph at 58X (b) Crack Locus Compared to that of Tyson & Davies[5]

shade the other. At that point, the larger crack dominates and the smaller crack does not have sufficient stress at the tip to continue.

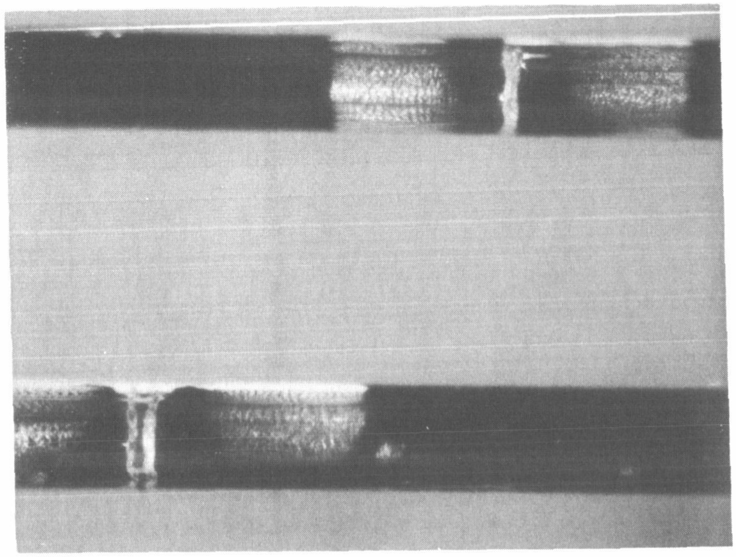
When the matrix is quite ductile and capable of resisting the tendency to crack initially, a bond failure may occur at the interface. This shear failure at the interface occurs near the filament end but not immediately at the end. The light areas in Fig. 6a represent the unbonded region. Schuster and Scala[4] have shown that the shear stress along the interface does drop off near the end of the filament as shown in Fig. 6b. One possible reason for increased bond strength near the end may be the radial contact pressure at the filament ends due to a possible vacuum in the fracture volume. This increased radial pressure should provide better shear strength just as increased normal force provides greater friction force.

To summarize the failure sources resulting from filament fracture, one can construct the model shown in Fig. 7. As the filament is loaded through shear at the interface, it may do one of three things:

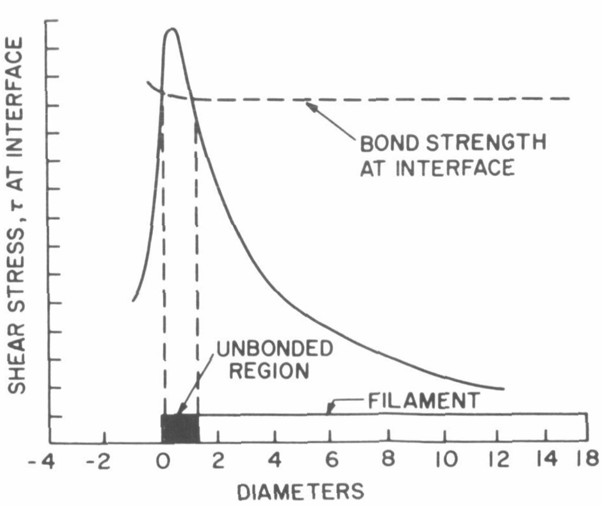
(1) Having sufficient bond strength and matrix tensile strength, the filament fractures at point A. This may result in a disk-shaped crack D, two inclined conical cracks C, both, or neither.

(2) If the bond strength is not sufficient to load the filament to fracture, it may simply unzip at the interface B. The shear stress pattern shifts along the interface toward the center of the filament and the process continues until no force is transferred to the filament other than friction at the unbonded interface.

(3) If the bond strength is sufficient to transfer load to the filament and if the matrix is weak in tension, a tensile crack C may propagate in the high shear transfer region before the filament fractures. This crack can then propagate becoming more normal to the filament as it leaves the region of high shear stress in the matrix.



(a)



(b)

Figure 6. Localized Bond Failure Occurring Near Filament Fracture
(a) Photograph at 116X (b) After Schuster and Scala(4)

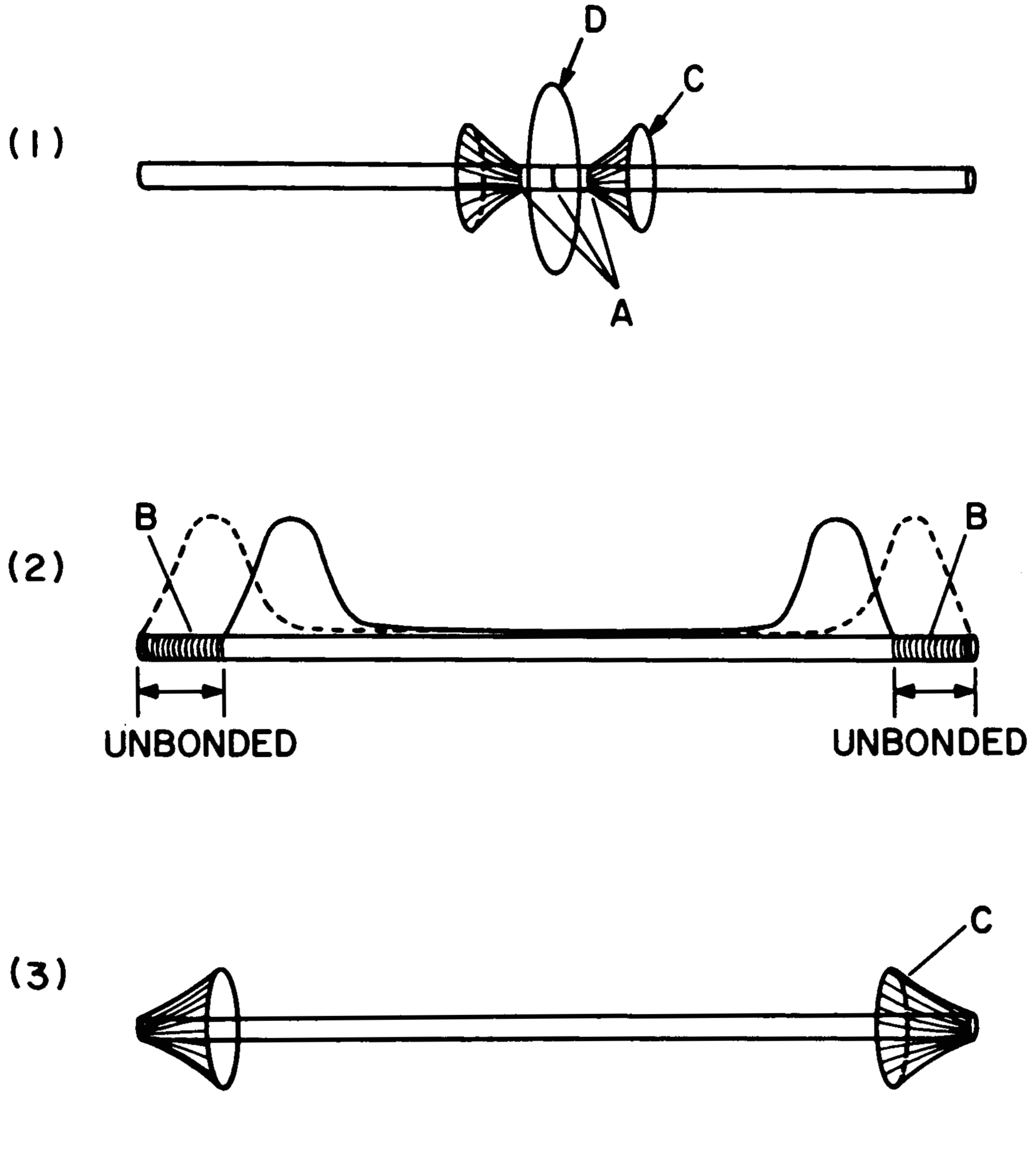


Figure 7. Summary of Failure Modes in a Single Filament

III. THE EFFECT OF IMPOSED STRAIN RATE ON THE LOCAL FRACTURE MODE AND MECHANICAL BEHAVIOR

Tensile tests were made on single-filament-epoxy specimens (Fig. 8) at different strain rates, and the results demonstrate that certain dynamic aspects of the (previously discussed) matrix fracture phenomena can have a profound effect on mechanical behavior.

One pair of (duplicate) specimens was pre-strained at a nominal strain rate of 0.008 in/in/min until at the maximum load (of the load-deflection curve) there was 10-13% elongation. These specimens were then unloaded for examination under the microscope, and the number of filament breaks was determined to be 12 in one case and 13 in the other (Table I). The pre-strained (cracked) specimens and, for comparison, a pair of virgin specimens were then tested to failure at a 100-fold higher strain rate (0.8 in/in/min). Representative stress-elongation curves (at the higher strain rate) for pre-strained and virgin specimens are shown in Fig. 9. The critical observation here (evident from Fig. 9 and Table I) is that, at the higher strain rate, the specimens with many internal cracks already present were both stronger and more ductile than virgin specimens (twice the strength and three times the elongation to failure).

The formation of two kinds of matrix tensile cracks (disk-shaped and inclined) which emanate from a filament tensile crack has been previously described in detail. Both of these kinds of cracks were present in the tensile test specimens, but the number of each and the extent to which each kind grows, can be seen to be (Table I and Figs. 10 and 11) dependent on the strain rate and/or the total elongation. The uncracked (virgin) specimen developed only a few filament fractures at the high strain rate before failure occurred. It appears from this that at high strain rate an unbroken filament can be a more likely source of catastrophic composite failure than an existing matrix crack. Furthermore, the specimens which had cracks present after loading at low strain rate did not develop additional cracks when loaded to

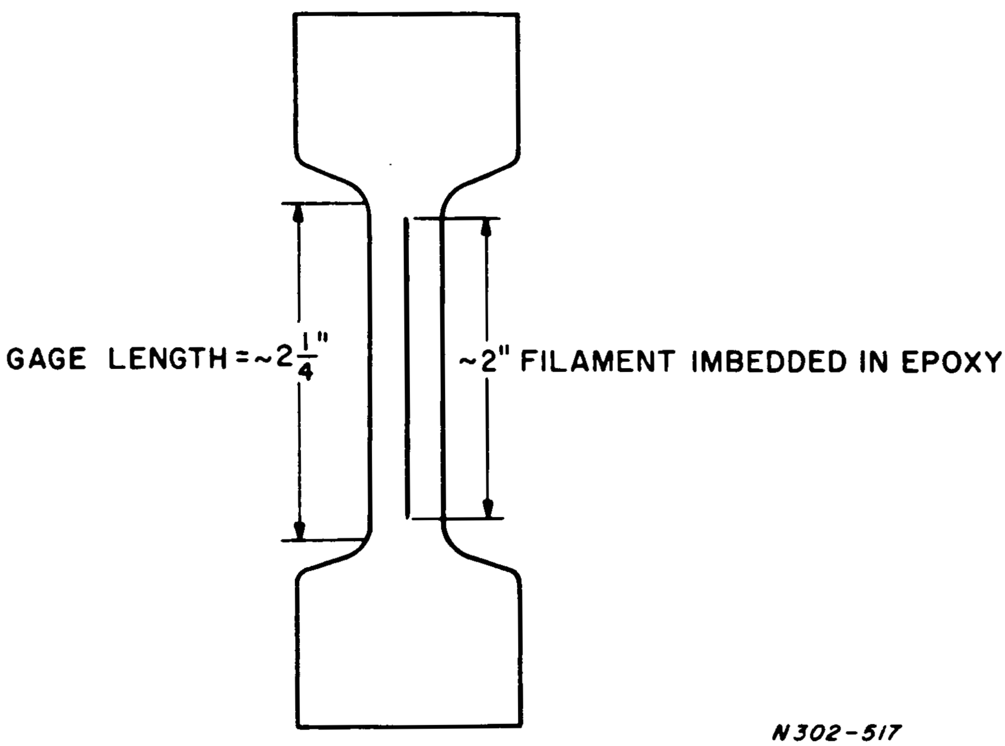


Figure 8. Schematic Representation of a Single Filament Unreinforced Composite.

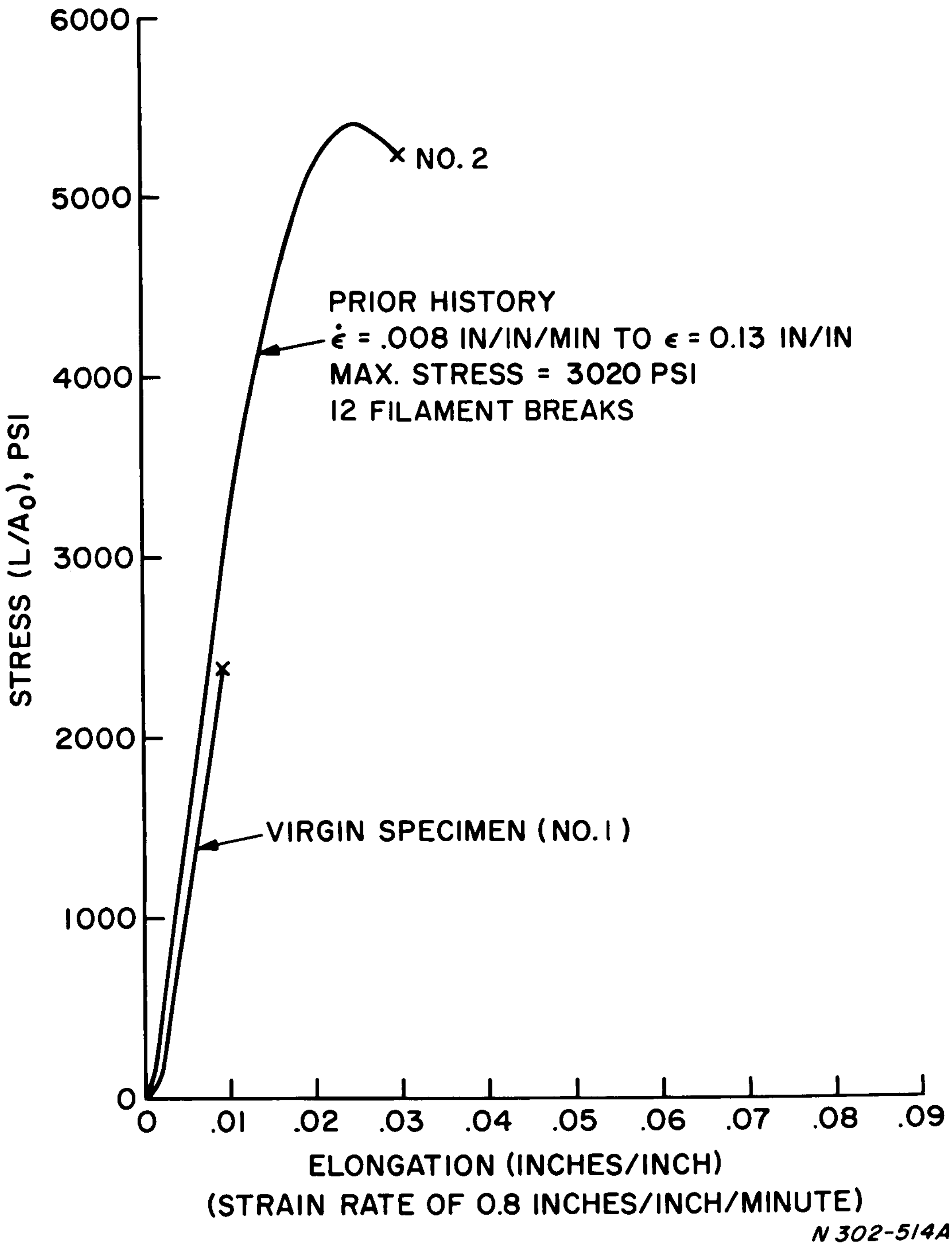


Figure 9. Stress-strain Behavior of Virgin and Pre-strained Epoxy Specimens Strained at 0.8 in/in/minute.

TABLE I

Tensile Data from Epoxy-B/W/Single Filament Composites

Specimen No.	Strain Rate (in/in/min)	Stress at Max. (Load, psi[1])	Total Strain (at Max. Load, in/in[2])	No. of Filament Breaks
1	0.8	2380[3]	0.009	7[5]
1A	0.8	2550[3]	0.010	4[5]
2	0.008	3020[4]	0.130[6]	12
	0.8	5380[3]	0.03	12[5]
2A	0.008	2800[4]	0.100	13
	0.8	5600[3]	0.04	14[5]
Plain Epoxy	0.8	(7)	(7)	(7)

(1) Load/original area
(2) Cross Head Travel/specimen gage length
(3) Fractured at maximum load
(4) Test stopped, specimen examined for cracks in filament and test resumed at high strain rate
(5) Including break associated with final fracture
(6) Almost all of this chiefly inelastic deformation was recovered soon after unloading
(7) The load exceeded the range being used and the strength was estimated to be between 6000 and 8000 psi with much more elongation than others at $\dot{\epsilon} = 0.8$ in/in/min.

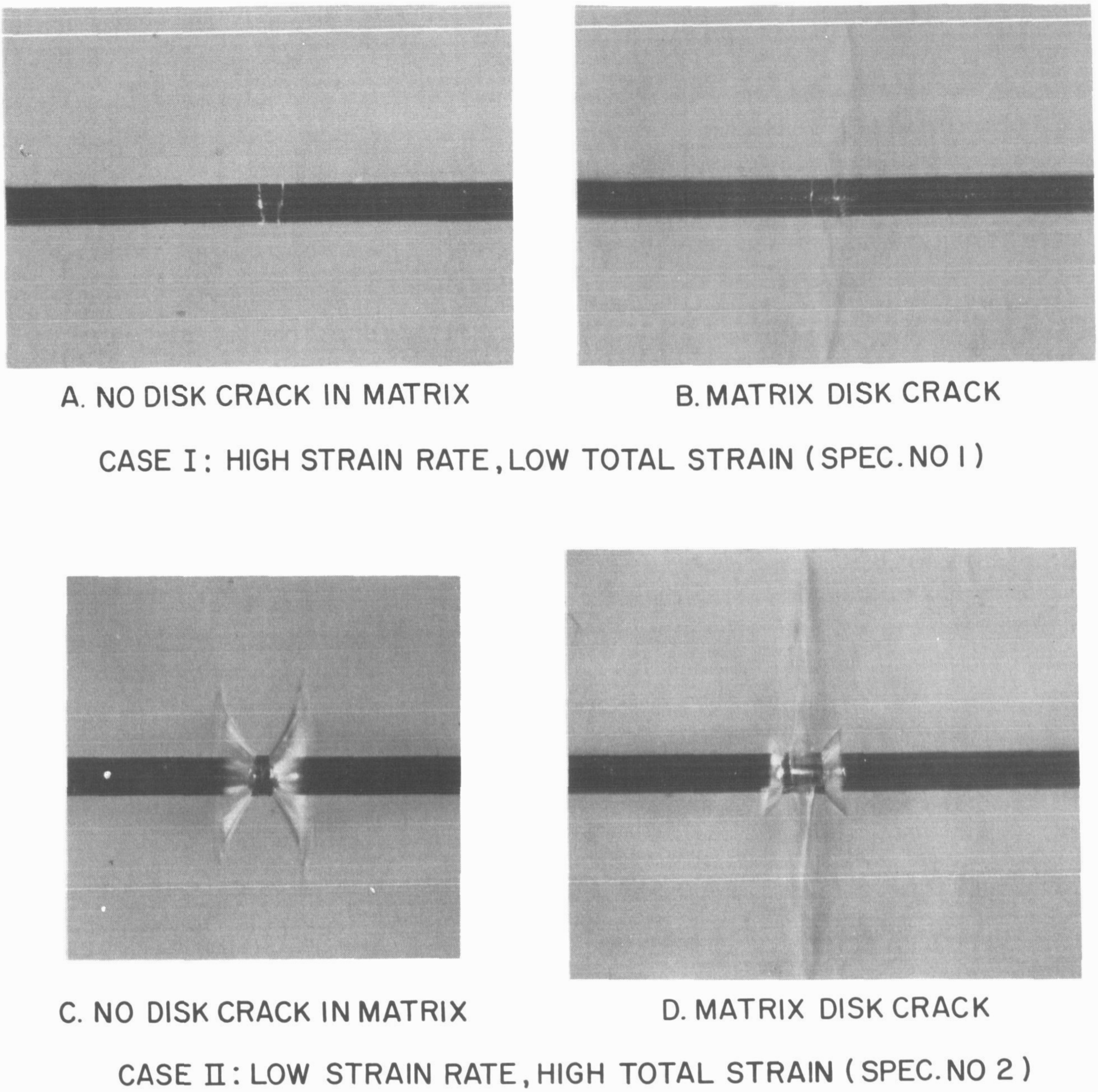
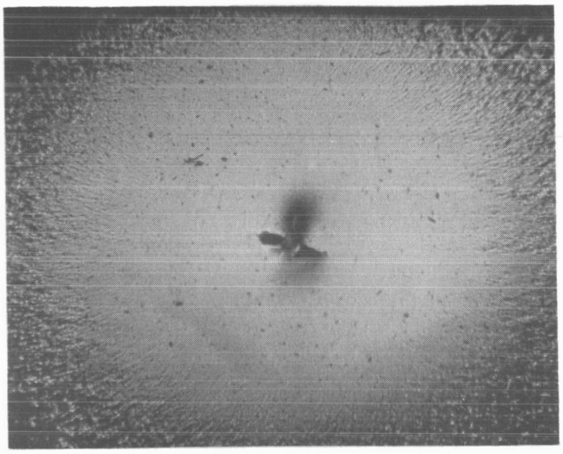
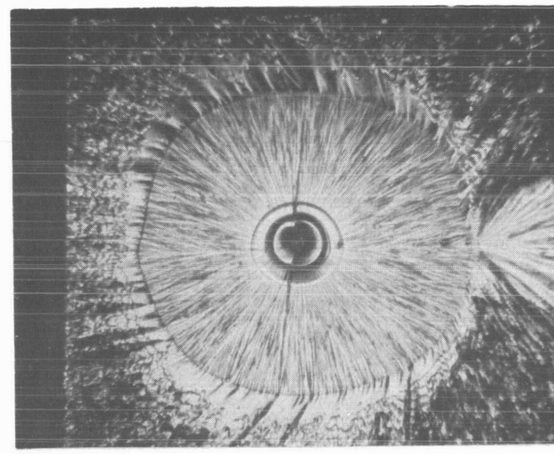


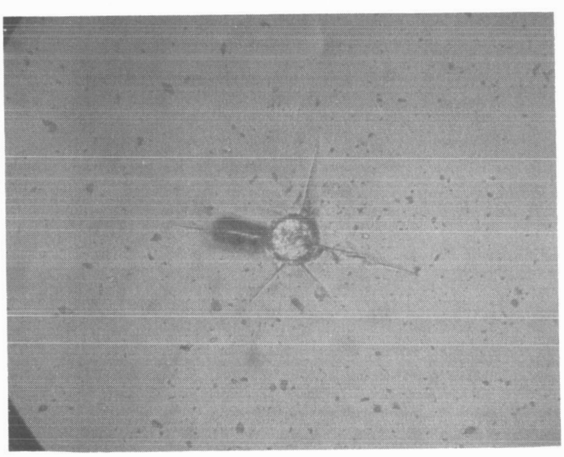
Figure 10. Two Consequences of Disk Cracks in B/W Filaments in Single Filament Epoxy Composites (Oblique Illumination, Longitudinal Views, 58X).



A. SPECIMEN NO. I (17X)

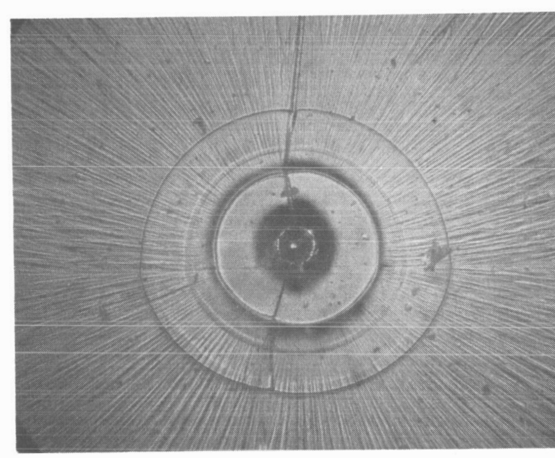


C. SPECIMEN NO. 2 (17X)



B. SPECIMEN NO. I (58X)

CASE I: STRAIN RATE = 0.8 IN/IN/MIN.
TENSILE STRENGTH = 2380 psi
AT ϵ_L = 0.009 IN/IN.
TOTAL OF 7 BREAKS



D. SPECIMEN NO. 2 (58X)

CASE II: STRAIN RATE 0.008 IN/IN/MIN.
CRACKS FORM FROM 1200 psi
TO 3020 psi. TOTAL OF 12
BREAKS AT STRAIN RATE OF
0.8 IN/IN/MIN, TENSILE
STRENGTH = 5380 psi AT
ϵ_L = 0.03 IN/IN.

Figure 11. The Effect of Strain Rate History on the Fracture Mode of B/W Filaments in Epoxy (Single Filament Composite).

fracture at the higher strain rate. That is, rather than developing new filament fractures, the cracks already present propagated to cause gross failure. It appears from these phenomena that the existing cracks behaved like energy sinks and were capable of absorbing a good deal of energy before reaching critical size.

The Photomicrographs on the left hand side of Figure 10 show the development (in the absence of a disk-shaped crack) of a pair of inclined cracks in the matrix associated with a low-energy filament failure. Figure 10A shows the initial stage of such a pair of cracks which, owing to catastrophic failure elsewhere at a low total elongation, did not have time to develop to the extent of the pair shown in Figure 10C, grown at a lower strain rate. The inclined cracks in Figure 10D were partially shadowed from the tensile stress field in the matrix (by the disk-shaped crack) and, therefore, they did not grow to the extent of the unshadowed pair in Figure 10C.

The photomicrographs in Figure 10B and D show the profiles of typical non-catastrophic disk-shaped tensile cracks originating at failures in the higher strength portion of the filament. In these specimens, such cracks had a minimum size of about 0.03 inch diameter which indicates that, while forming, they grow very rapidly. The new surface produced in this initial stage of crack propogation is, by virtue of the intense stress concentration and high speed, very smooth. This smoothness is evident for both catastrophic (Figure 11, A and B) and for first stages of non-catastrophic cracks (Figure 11, C and D). After the sudden energy release has been absorbed in creating the smooth fracture surface, non-catastrophic cracks slow down and the rate of propagation is eventually governed by the crack configuration and the externally imposed strain rate. The more slowly moving crack tip is preceded by some inelastic deformation and can propagate on slightly different levels when it is energetically advantageous to do so. This stage of crack propagation is characterized by the radial markings evident in Figure 11, C and D. Such cracks continue to grow until they become super-critical in size and self-propagating; that is, the specimen fails under conditions of decreasing load (as shown in the upper stress strain curve in Figure 9).

In summary, a transient energy release at a filament fracture can result in a smooth high-speed crack in an epoxy matrix. Other test conditions and matrix properties determine whether or not such a crack will propagate catastrophically. The phenomenology (and consequence) of filament failure might be similar or quite different in a metal matrix. However, even if a metal matrix was not damaged in any obvious way, the undamped remnant of a stress pulse would be transmitted to nearby filaments in a manner not adequately described by a mere redistribution of static loads.

IV. MULTIFILAMENT COMPOSITE BEHAVIOR

Now that the response of the matrix to a single filament fracture has been described, the observed phenomena can be related to multifilament composite behavior. Consider several high modulus filaments in the same plane reinforcing a matrix of lesser modulus. Schuster and Scala[5] have shown that for filament spacing greater than 5 or 6 diameters there is no appreciable interaction of the stress fields at the ends of the filaments. When spacing is less than this the stress transfer between filament and matrix is concentrated in a region nearer to the filament and there is considerable interaction in that region. The shear stress concentration at close spacing approaches 10 (ratio of τ_{max} to $\sigma_c/2$) when two filament ends are adjacent. Schuster and Scala[5] point out that staggering the ends in short filaments reduces the shear stress concentrations to levels below that occurring in single filament composites. This is due to reinforcement of the matrix by the adjacent filament. The resulting stress field at a filament break between two other filaments will then differ from that of a single filament specimen. When such a fracture occurs with no disk shaped crack forming in the matrix, the tensile stress trajectories might appear as shown in Figure 12a.

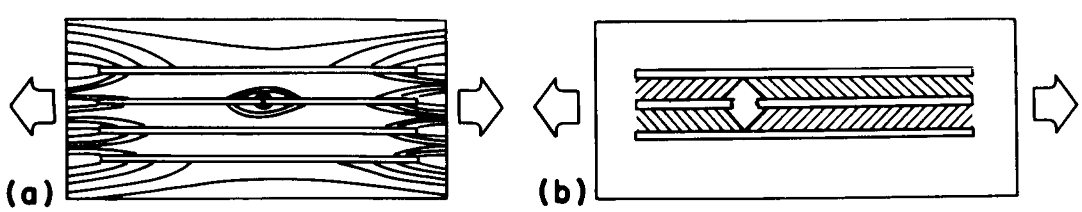


Figure 12. Redistribution of Load at Filament Fracture Site

Some stress transfer occurs across the break through the matrix, but because the adjacent filaments provide greater stiffness than the unreinforced matrix, there is shear transfer through the matrix to the adjacent filaments. This mechanism is shown in Figure 12b and becomes more pronounced as filament spacing decreases. In this sketch one might consider the broken filament as being bonded between the adjacent filaments as in a standard pull-out test of a double lap joint. Through this mechanism, much of the load previously carried by the fractured filament is now shared by unbroken filaments, especially those adjacent to the broken filament. Since the direct stress transfer through the matrix (Figure 12a) requires that the matrix remain intact near the fracture, a disk crack at the filament fracture will prevent stress transfer through the resin between filament ends. Under this condition, all the stress transfer must be through shear to adjacent filaments as illustrated in Figure 12b.

When such a disk is formed with filaments closely spaced, enough energy may be released to carry the disk-shaped crack through the adjacent filaments resulting in a multi-filament fracture and premature failure of the composite. There is evidence that this mode of failure is strongly dependent on strain rate. Two specimens were prepared with 5 filaments in an epoxy resin. One was loaded at a strain rate of 0.008 and the other at 1.0 in/in/min. At the slow rate, individual filament fractures occurred in various locations throughout the test and several subcritical disk-shaped cracks were formed as the stress was increased to a maximum of 2180 psi. In the specimen tested at the higher strain rate, only three filament fractures were sustained before total failure at a stress of 1150 psi. A diagram of the fractures occurring in these tests and two single filament tests are shown in Figure 13. The slower rate of loading allows gradual redistribution of load and failure of filaments according to their variations in strength from point to point. At the higher strain rate a single filament fracture propagates more rapidly through the matrix and therefore adjacent filaments are loaded rapidly. The result

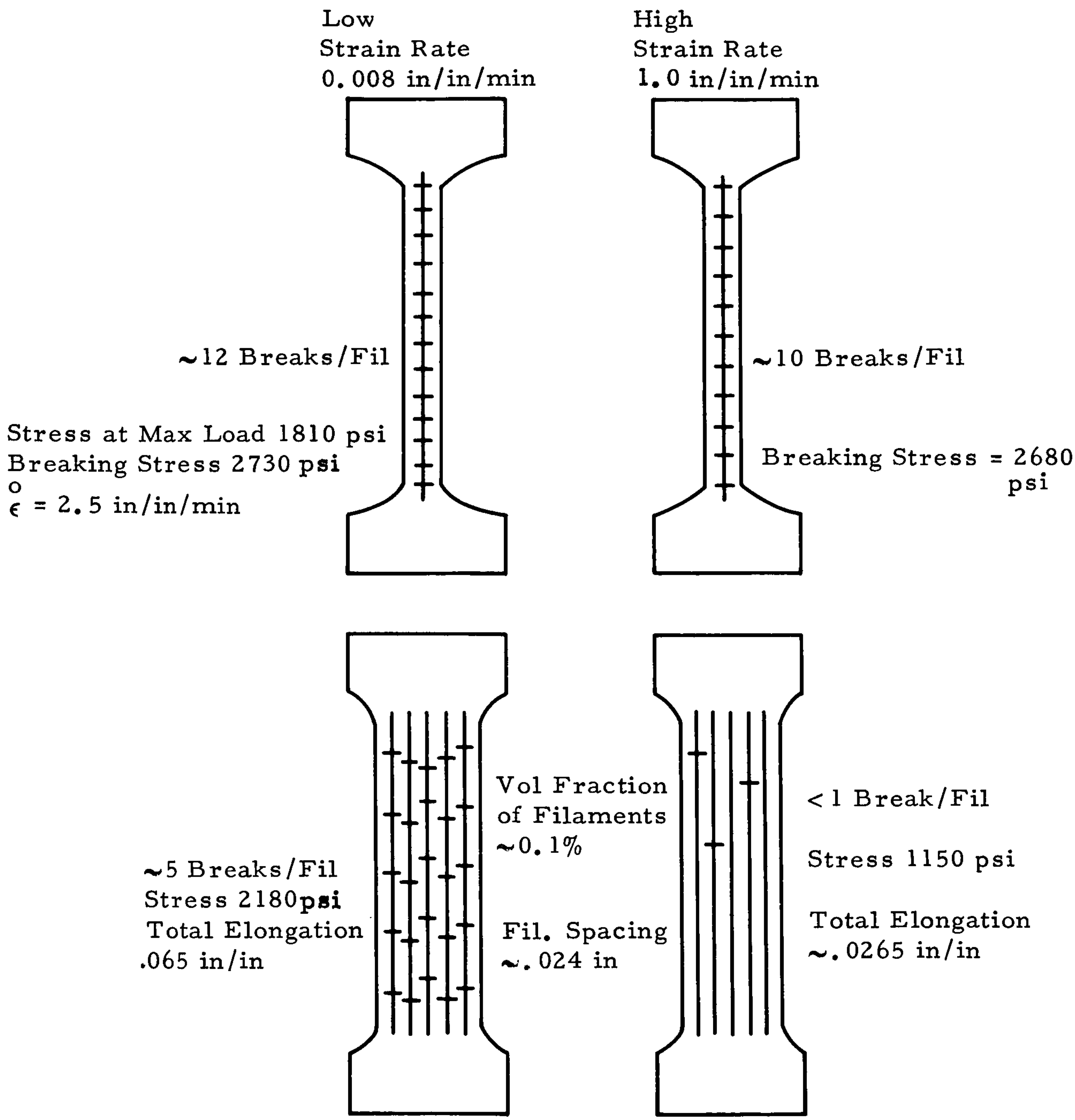


Figure 13. The Effect of Strain Rate and Filament Spacing on the Number of Breaks/Filament

is a rapid sequence of high-energy filament fractures beginning from one disk shaped crack. From these data it is clear that the influence of matrix cracking on composite failure increases with strain rate, and this is not accounted for in rule of mixture predictions. While it is difficult to take it into quantitative account, it would seem reasonable to suppose that the critical strain rate is that at the tip of the propagating matrix crack. This would be governed by the amount of energy and its rate of release during filament fracture and upon the strain rate sensitivity of the matrix.

For sometime a good deal of effort has been applied to improving the bond strength between matrix and filaments with the intention of obtaining better stress transfer at the interface. At first glance, this appears to be a logical step toward improving the reinforcing mechanism and therefore developing higher strength in composites. Certainly, the bond strength must be sufficient to load the filaments to their fracture strength. However, if the bond strength is greater than that needed to fracture the filament it will cause the matrix to fail in tension at the filament fracture site. This matrix crack then causes premature fracture of adjacent filaments and the gross strength of the composite may, in fact, be reduced. It would be far more desirable to provide just enough bond strength to fracture the average strength filaments and no more. At this level of bond strength, the matrix would unbond at the newly formed ends rather than fracturing normal to the tensile stress trajectories. Figure 14 shows such a failure process with the detail of a single fracture shown previously in Figure 6A. When the bond strength and the tensile strength of the matrix are very nearly matched, the optimum condition results. Each time an inclined matrix crack begins, the filament unbonds and the inclined crack is retarded. This results in an unbonded surface intermittently punctuated with inclined cracks which have been arrested. This unbonded surface is illustrated in the bottom filament, Figure 15. Note that the filament in the center produced a disk crack and then unbonded.

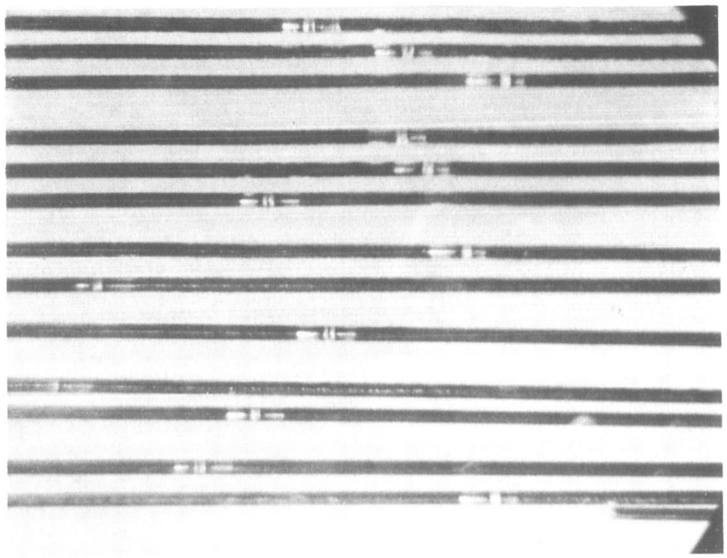


Figure 14. Patterns Typical of Bond Failure at Filament Fractures (17X)

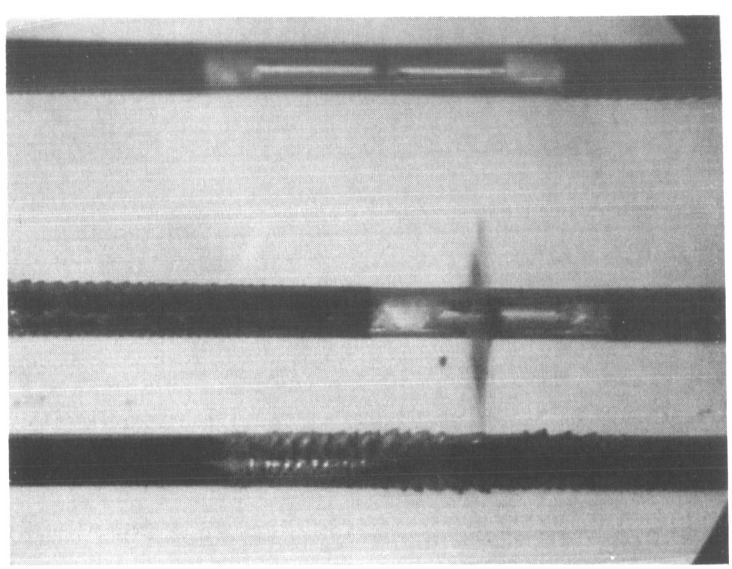


Figure 15. Unbonded Region Where Tensile Failure Has Been Retarded (58X)

In most of the figures presented on previous pages, the photographs have been taken after load removal. To better appreciate the crack geometry under loading, Figure 16 illustrates several different fracture modes occurring in two specimens while the load is applied. Note that cracks open considerably during load application into configurations more descriptive of the stress trajectories alluded to previously. Upon release of the load the cracks close and the geometry is changed considerably. Figure 16 shows four separate filament fractures, each with a different matrix crack pattern. The first photo (#A) shows two inclined matrix cracks with no accompanying disk shaped crack. Photo (#B) shows the beginning of a disk shaped crack. Photo #C shows a larger single disk in the open position. The geometry gives a more accurate estimate of the stress trajectories of transfer around the crack. Photos D & E show both disk and inclined cracks in an open position. In these photos it is obvious that the inclined cracks are not shaded by the disk to nearly the degree that would appear in the unloaded condition. All the photos are illuminated with polarized light, but no attempt was made here to analyze the specimens photoelastically.

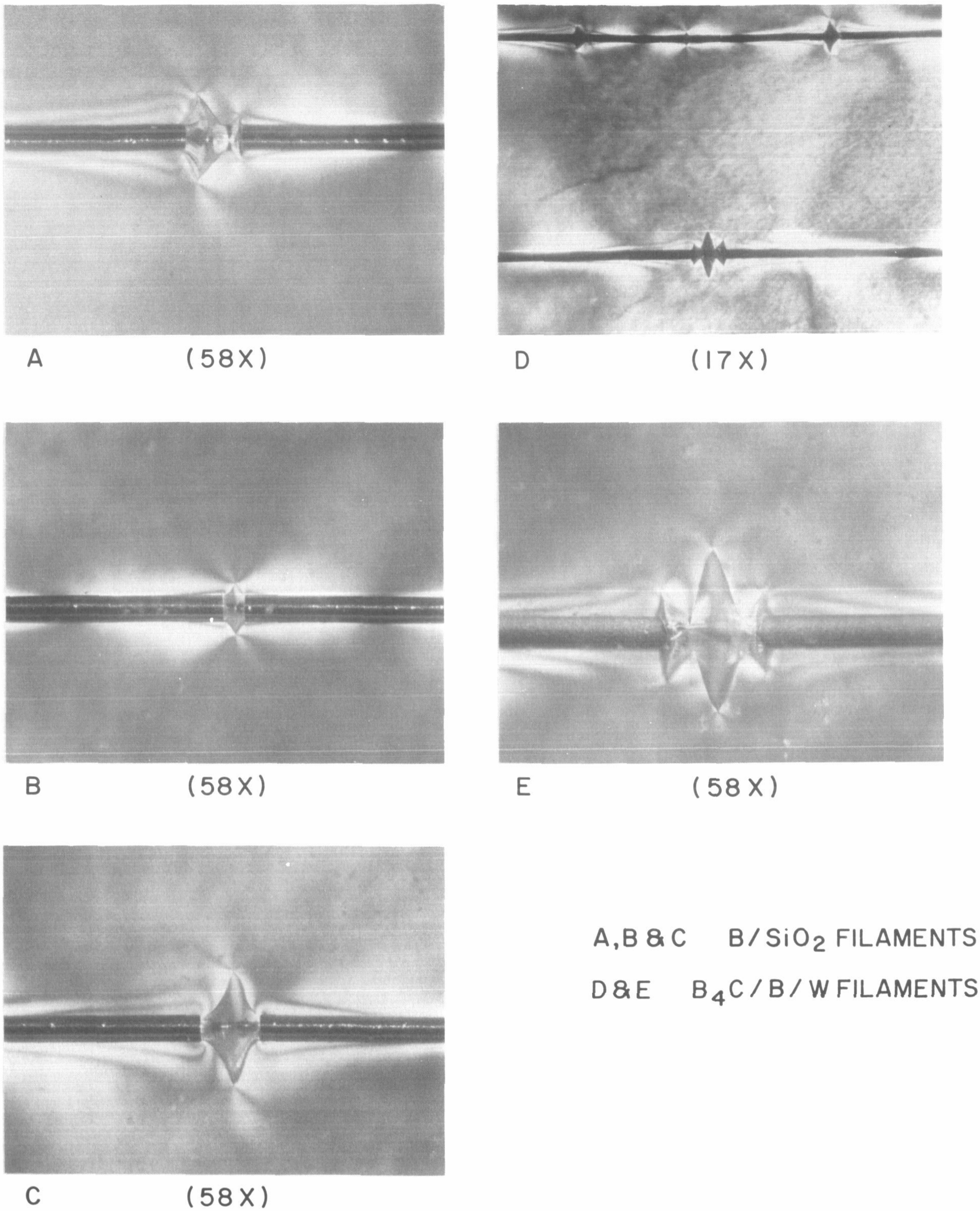


Figure 16. Photographs of Filament Fractures Under Load.

V. CONCLUDING REMARKS

It has been established that there are certain distinct modes of failure of the reinforced epoxy specimens in the vicinity of a filament fracture. Two of these failure modes involve crack propagation in the matrix and the third is unbonding at the interface between matrix and filament. Of the three modes of failure, the least damaging from the standpoint of overall composite integrity is the unbonding mechanism since it does not immediately cause fracture in adjacent filaments. It is significant that the unbonding mechanism generally yields better agreement with rule of mixture predictions. Glass filament reinforced epoxy specimens exhibit this mode of failure and have been used by Rosen[(5)] to study the failure process.

It is interesting to note that rule of mixtures predictions for strength of resin matrix composites attribute very little of the composite strength to the matrix. That is, the matrix contribution ($V_m \sigma_m$) is usually quite small compared to the filament contribution ($V_f \sigma_f$) and is sometimes neglected when the volume fraction of filaments is high. This approach presumes that the matrix cannot detract from the composite strength, and this is not necessarily true.

A composite can be either filament limited or matrix limited when there is sufficient bonding to develop the fracture strength of the filaments. Studies aimed at improving composite strength generally focus on using very strong filaments and developing the best possible bond between matrix and filament. Certainly a good bond is essential to the reinforcing mechanism, but too strong a bond may render the composite severely matrix limited.

Consider a matrix reinforced with filaments having some statistical distribution of filament strengths. When the weaker filaments are loaded to their fracture strengths, a very strong bond will result in matrix cracking more or less normal to the filament. In propagating to adjacent filaments which have not yet reached their fracture strength, the matrix cracks can prevent the stronger filaments from developing their potential strength. The overall strength of the composite is therefore reduced. This condition becomes more critical when filaments are closely spaced, when the matrix is quite crack sensitive or when the load is rapidly applied. Were the bond strength not so great it might not develop the strength of the strongest filaments, but this seems a reasonable price to pay to develop the strength of most of the filaments. With lower bond strength the matrix would tend to unbond at the filament fracture sites and matrix cracking would be minimized. Whatever the bond strength, the source of a matrix crack is the sudden release of energy from the filament fracture.

This study shows that the response of epoxy to sudden release of energy results in disk-shaped matrix cracks. Analagous behavior in metal matrices may result in local plastic deformation rather than crack initiation, but the transient response would involve transmission of a stress pulse to adjacent filaments. Though some attenuation of this pulse would occur, there is the likelihood that this sudden loading of adjacent filaments exceeds that predicted by simply redistributing the static load.

A step toward alleviating this condition is to reduce the sudden energy release mechanism. A single test was conducted with a tungsten filament as the single reinforcing filament. This filament was reported to have negligible ductility as measured in a standard tensile test but necked considerably when loaded in the matrix (See Figure 17). The flow type failure of the tungsten filament prevented the sudden release of energy normally encountered at fracture of a boron or silicon carbide filament. This suggests that a filament of boron on a ductile substrate could deform sufficiently to retard a sudden energy release. Boron on silica might provide an analogous capability by virtue of the low modulus of the silica core.

Another means of retarding matrix crack propagation might involve placing ductile filaments at intervals between boron or silicon carbide filaments. When a crack reached the ductile filament it would then be arrested rather than causing fracture of that filament and further release of energy. This approach does not attempt to eliminate the source of the matrix crack and therefore might be augmented by controlling the bond strength to minimize matrix cracking.

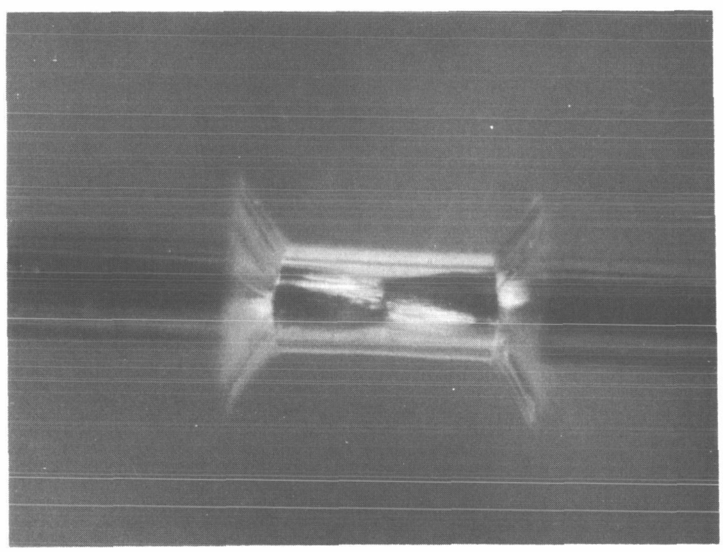


Figure 17. Example of the Necking and Simultaneous Debonding of Tungsten Wire in an Epoxy Matrix (116X)

Perhaps an even more fundamental aspect of this study is the identification of fracture phenomena which allow reconstruction of the fracture process. Disk-shaped cracks suggest a high energy fracture while the absence of a disk suggests low energy fracture, perhaps at a weak spot in the filament. The growth of inclined cracks near newly formed ends represents a more gradual fracture process and is a clue to the chronology of the fracture process. The presence of a smooth disk with no inclined cracks would then indicate a high energy fracture late in the failure process.

Finally, the single filament test provides a useful tool for evaluating both matrix and interface characteristics. Knowing the tensile strength of the matrix one can compare bond strength to matrix tensile strength by observing which mechanism takes precedence in the loaded specimen. This approach makes possible the comparison of resin properties, filament surface characteristics and surface treatments under actual conditions of interface behavior in a relatively simple and inexpensive test.

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SUMMARY The object of this study is to establish experimentally the critical fracure modes of an epoxy matrix in the vicinity of a break in a high modulus, high strength filament. Three distinct failure modes were observed to occur and the nature of these three modes is explained through an analysis of the stress state in the matrix. A single filament embedded in epoxy novolac was used to examine the fracture process. The advantage of using this type of resin arises from its ability to be modified to varying degrees of crack sensitivity.

The effects of varying gross strain rate were evaluated and the study was then extended to specimen configurations containing more than one filament. This approach permits the examination of the interaction of nearby filaments with localized fracture processes. Filaments of boron on tungsten (B/W) boron on Silica (B/SiO_2), boron carbide on B/W and tungsten wire were used with epoxy novolac (DEN 438).

KEY WORDS Composites, Behavior, Fracture Phenomenon, Stress Analysis

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