Mechanisms Controlling Fatigue Damage Development in Continuous Fiber Reinforced Metal Matrix Composites

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ABSTRACT

Damage in continuous fiber reinforced metal matrix composite materials can be quite complex since there are a number of different constituents (fiber, matrix, and the fiber/matrix interface) that can fail. Multidirectional lay-ups have an even greater number of possible damage orientations and mechanisms. Based on the simplifying assumption of equivalent constituent strain states in the absense of damage, a strain based failure criteria may be applied to determine when and where initial damage will occur. Based on the relative strain to fatigue failure of the fiber and matrix, the possible damage mechanisms of an MMC can be grouped into three categories: (1) matrix dominated, (2) fiber dominated, and (3) self-similar damage growth. A fourth type of damage development, fiber/matrix interface failure, is dependent on the relative strength of the fiber/matrix interface and the matrix yield strength. These four types of damage are discussed and illustrated by examples. The emphasis is on the fatigue of unnotched laminates.

KEYWORDS

Fatigue; continuous fibers; metal matrix composites; aluminum; titanium; silicon-carbide fibers; interface.

INTRODUCTION

Continuous fiber reinforced metal matrix composites (MMC) are projected for use in high temperature, stiffness critical parts that will be subjected to cyclic loadings. Fatigue of metal matrix composites can be quite complex. A metal matrix, because of its relatively high strength and stiffness compared to the fiber, plays a very active role compared to a polymer matrix. For example, Johnson (1982) showed that fatigue damage in the matrix of boron/aluminum can reduce the laminate stiffness by as much as 50% without causing laminate failure.

The fatigue failure modes in continuous fiber reinforced metal matrix composites are controlled by the three constituents of the system: fiber, matrix, and fiber/matrix interface. The relative strains to fatigue failure of the fiber and matrix will determine the failure mode provided the fiber/matrix interface is strong enough to support the load. Based on the simplifying assumption of equivalent constituent strain states in the absense of damage, a strain based failure criteria may be applied to determine when and where initial damage will occur. If the matrix requires much less cyclic strain to initiate fatigue damage than does the fiber, then the damage will be matrix dominated. If, on the other hand, the fiber requires less cyclic strain to fail than does the matrix, the damage will be fiber dominated. A fiber damage dominated composite will fail rather suddenly in fatigue with little warning, provided the fiber/matrix interfaces do not fail. If both the fiber and matrix require approximately the same cyclic strain for fatigue failure and the fiber/matrix interface is sufficiently strong, self-similar crack growth, as found in metals, may result. Self-similar crack growth is also possible when the matrix is strong enough to create a high stress concentration in the fiber ahead of a matrix crack. Thus, by starting the fatigue damage in the matrix, the crack can propagate across the fibers. The fiber/matrix interface may be unable to carry the required stress and may fail, causing fiber/matrix separation. This is likely to occur in those MMC systems with high yield strength matrices that require high load transfer between fiber and matrix in the off-axis plies.

As new continuous fiber-reinforced metal matrix composites are hypothesized and developed, projections of their fatigue behavior can be made by understanding the relative strengths of the fiber, matrix, and the fiber/matrix interface. This paper is a short review of the author's past work, intended to illustrate some of the controlling damage mechanisms in continuous fiber reinforced metal matrix composites.

FAILURE MECHANISMS

Based on the relative strain to fatigue failure of the fiber and matrix, and the interface properties, the possible failure mechanisms of MMC can be grouped into four categories: (1) matrix dominated, (2) fiber dominated, (3) self-similar damage growth, and (4) fiber/matrix interfacial failure. These four types of damage will be discussed and illustrated by examples in the remainder of the paper.

Matrix Dominated Damage

In boron/aluminum (B/Al) composites, the boron fibers are very fatigue insensitive. They are rather large diameter (0.14 mm) fibers, with very smooth sides, and are virtually elastic until fracture. Boron has a strain to static failure of about 0.0085. While the strain to failure of 6061-0 aluminum is almost 0.1, it yields at a strain between 0.001 and 0.002. Therefore, under static loading the fibers would reach their critical strain before the matrix as shown by Johnson, Bigelow and Bahei-El-Din (1983). But under fatigue loading, the matrix would cyclically yield at strain levels far below the critical strain of the fiber. This cyclic yielding could result in fatigue damage to the matrix but would not affect the fibers. Thus in this composite, the initial fatigue damage is matrix dominated.

First, let us examine why fatigue cracks would develop in the matrix material. If fatigue damage in general is to be avoided, and low cycle

fatigue failures in particular, the cyclic loading must produce only elastic strains in the constituents. Even so, local plastic straining can be permitted in the composite during the first few load cycles, provided that the composite "shakes down" during these few cycles. The shakedown stress is reached if the matrix cyclically hardens to a cyclic yield stress Y such that only elastic deformation occurs under subsequent load cycles. The value of Y is 140 MPa for annealed 6061 aluminum (Johnson (1979)). This implies a cyclic strain range of approximately 0.002 for the onset of fatigue damage in the aluminum matrix. Bahei-El-Din and Dvorak (1982) formulated an analysis to calculate the yield surfaces of any laminate. This analysis was used by Johnson (1982 and 1983) to calculate the shakedown range for aluminum matrix composites. The possible relationship between fatigue and shakedown in metal matrix composites was first suggested by Dvorak and Tarn (1975) and related to then available experimental data, obtained primarily for older unidirectional 6061 B/Al materials. Dvorak and Tarn suggested that the shakedown stress was related to laminate fatigue failure. In subsequent research (Dvorak and Johnson (1980) and Johnson (1982 and 1983)), using newer specimens and more detailed testing, the relationship was examined theoretically and experimentally for both unidirectional and cross-plied laminated 6061-0 B/Al composites. They found that the shakedown stress was related to the onset of fatigue damage in the composite matrix material but not to laminate failure (failure of the load carrying 0^{o} fibers).

Next, an example will be given of the type of fatigue damage that occurs in B/Al and silicon-carbide/aluminum (SCS_2/Al) composites (Johnson (1982) and Johnson and Wallis (1986), resp.) containing 0^{o} plies subjected to cyclic loadings above their respective shakedown range. Figure 1 presents the fatigue damage accumulation as a function of number of applied cycles and stress for a B/Al $[0/90]_{2s}$ laminate. The damage is expressed in terms of E_{ul}/E_I, which is the ratio of the elastic unloading modulus remaining after N number of cycles to the initial elastic modulus given as a percent. All of the data is for specimens that survived at least 2,000,000 cycles, after which the tests were terminated. Notice that each specimen appears to reach a stabilized value of E_{ul}/E_I, herein referred to as a "saturation damage state."

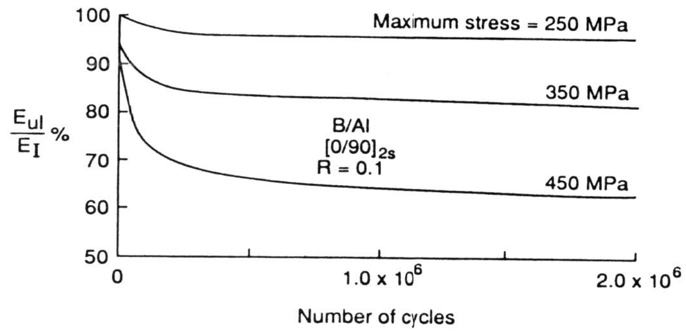

Fig.1. Percentage of initial elastic unloading modulus retained as a function of applied cycles.

Etching away the matrix material and exposing the fibers revealed that essentially no 0^{o} fiber breaks occurred unless the specimen was fatigue loaded to within 10 percent of the fatigue limit load (500 MPa at R= 0.1).

The drop in the elastic unloading modulus of those specimens cycled at stress levels below the fatigue limit can be attributed almost entirely to cracking in the matrix material between the fibers in the off-axis plies. Figure 2 shows the matrix cracks in the 90° layer just beneath the 0° ply of a B/Al laminate. The cracks in the matrix material are parallel to the 90° fibers. Pictures of other such damage can be found in Dvorak and Johnson (1980) and Johnson (1983). The matrix cracks in the off-axis plies did not grow past the 0° fibers to the surface of the specimen.

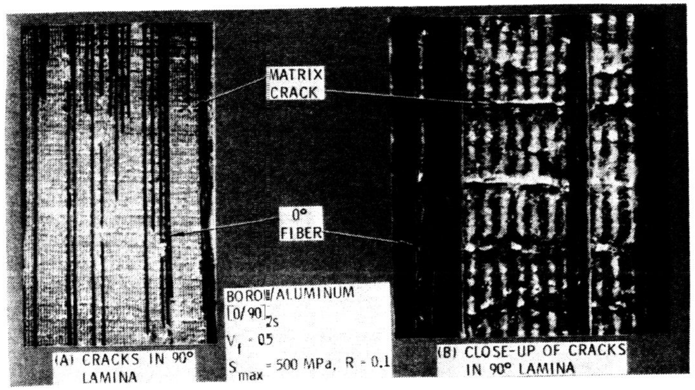

Fig. 2. Matrix cracks in the 90° lamina as exposed by etching away the matrix material in the 0° ply.

The author has formulated a model that predicts the effect of matrix cracking on laminate stiffness (Johnson (1983)). This model has been verified for a variety of different lay-ups of B/Al and SCS_2/Al composites.

Experimental results for matrix dominated fatigue damage indicate the existence of three distinct regions in the S-N plane in which one observes different responses of MMC to cyclic loading. Figure 3 illustrates these regions for a $[0/\pm45/90/0/\pm45/\overline{90}]_s$ B/Al laminate. At low stress levels, below the shakedown stress limit (218 MPa), there is no significant accumulation of fatigue damage. The elastic modulus and static strength remain intact up to, and probably beyond, two million cycles. Above the shakedown stress level there is a damage accumulation region, where reductions in the elastic modulus are observed after a certain number of cycles. The S-N curve is a boundary between the damage accumulation region (matrix cracking) and the fracture region (0° fibers failing).

If a designer was only concerned with the material S-N fatigue behavior shown in Fig 3, he would chose 70% of ultimate as a safe design load for a life of at least 2 million cycles. However, at that load the MMC would experience a significant loss of stiffness. If the designer wished to retain all of the initial stiffness for the 2 million cycle lifetime, then he should not allow the cyclic stress range to exceed 35% of ultimate for this particular composite.

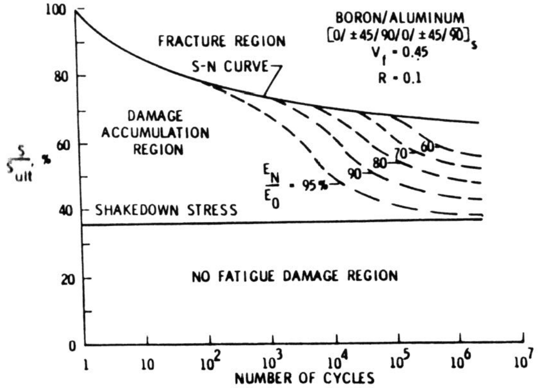

Fig. 3. Three regions of response to fatigue loading.

Fiber Dominated Damage

Recent work by Johnson (1989) on a silicon-carbide fiber reinforced titanium matrix composite (SCS_6/Ti-15-3) has shown that the stress level in the 0° fibers may be a governing parameter for predicting fatigue life. S-N data was experimentally determined for four different lay-ups containing 0° plies. The stress-strain response was monitored during the fatigue life. The stiffness dropped very early in the cycling history due to fiber/matrix interface failures. (This will be discussed in greater detail in a later section.) After a few cycles the stiffness stabilized and the cyclic strain range was recorded. This stabilized strain range was multiplied by the fiber modulus (400 GPa) to determine the cyclic stress range in the fiber. The number of cycles to failure was then plotted against the cyclic stress in the 0° fibers. The fatigue data from the four different laminates was correlated very well by the 0° fiber stress as shown in Fig. 4. Since the laminate will not fail until the 0° fiber fails, it is reasonable to assume that the stress in the 0° fiber will dictate fatigue life.

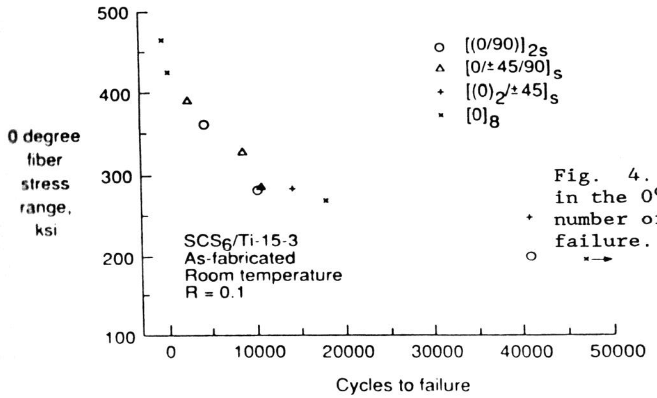

Fig. 4. Cyclic stress range in the 0° fibers versus number of cycles to laminate failure.

It is important to note that the fiber failures were not the first damage to occur in these composites. The first damage that caused significant modulus changes was the failure of the fiber/matrix interface in the off-axis plies.

Self-Similar Fatigue Damage Growth

The author (Johnson (1989)) has recently found evidence of self-similar crack growth in as-fabricated, unnotched unidirectional specimens of SCS_6/Ti-15-3. The fatigue data shown in Fig. 4 indicates that the SCS_6 fiber may have a fatigue limit around 1300 MPa. Since the fiber modulus is approximately 400 GPa, the fatigue limit strain range for the fiber is about 0.0033. Fatigue tests conducted by the author on Ti-15-3 material indicate a fatigue limit strain range of approximately 0.0036 at R=0.1. Thus the approximate fatigue limit strain ranges for the fiber and matrix are very nearly the same. Figure 5 presents a photograph of a portion of a unidirectional specimen failure surface. The specimen was cycled at a R=0.1 with a maximum stress of 690 MPa. The specimen failed at 519,000 cycles. Half of the failure surface was rather shiny, indicating the surface had failed in fatigue. The other half was rather dull, as is typical of a dimpled surface, indicating static failure. The portion of the failure surface shown in Fig. 5 lay within the fatigue area. Specifically, the photo shows where a semi-circular shaped crack intersected a larger crack that was growing from one edge of the specimen. This is a good illustration of damage initiating and growing as a self-similar crack. Although these composites have rather weak fiber/matrix interfaces, a combination of the residual radial compressive matrix stresses around the fiber and the Poisson's effect when pulling a unidirectional laminate in tension, resulted in interface strengths sufficient to allow self-similar crack growth in this laminates.

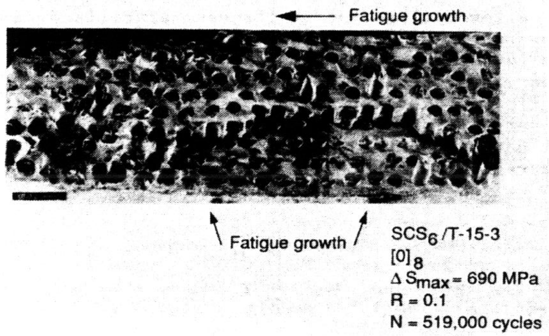

Fig. 5. Micrograph of the intersection of two fatigue cracks that have grown in a self similar fashion.

Fiber/Matrix Interface Failures

Recent work by Johnson, Lubowinski, Highsmith, Brewer, and Hoogstraten (1988) illustrates the effect of a weak fiber/matrix interface on the mechanical behavior of a titanium MMC (SCS_6/Ti-15-3). The initial

stress-strain curves for each laminate containing off-axis plies exhibited a knee in the loading curve at approximately 140 MPa, well below the matrix material's minimum yield strength of 690 MPa. In all cases, the unloading elastic modulus was also less than the initial elastic modulus, indicating that damage had occurred in the laminate. For cyclic loading after the first cycle, the unloading curve closely following the loading curve indicating an opening and closing phenomenon. This resulted in a bilinear response with a knee near 110 MPa. An edge replica technique confirmed that this phenomenon was due to fiber/matrix interface failures in the off-axis plies.

Figure 6 presents the elastic unloading modulus (E_{ul}) divided by the initial elastic modulus (E_I) versus the number of applied cycles for data from $[0/90]_{2s}$ laminates of boron/aluminum (B/Al) and the SCS_6/Ti-15-3. Both specimen were loaded such that the ratio of stress range to ultimate strength was approximately the same. In the case of the B/Al laminate, damage initiated in the matrix material of the off-axis plies and grew as a fatigue crack. Therefore, the B/Al stiffness did not drop during the cycles required for fatigue crack initiation. By 2 million cycles, the unloading modulus had dropped by almost 20 percent. This is in sharp contrast to the SCS^6/Ti-15-3 composite for which the modulus dropped more than 20 percent in the first cycle, then remained almost constant until just before failure at 10,000 cycles. This is a good example of how damage mode affects laminate stiffness.

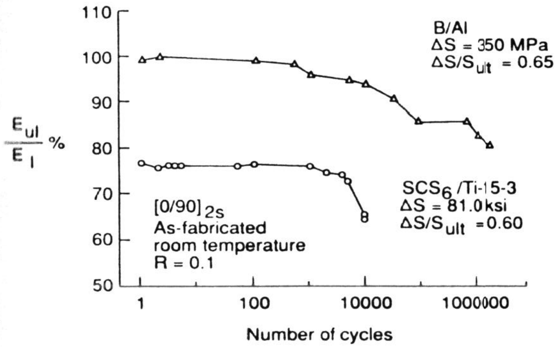

Fig. 6. Change in the elastic unloading modulus for $[0/90]_{2s}$ layups of B/Al and SCS_6/Ti-15-3.

Since the aluminum matrix composites did not have extensive fiber/matrix interface failures, the interfaces in the boron/aluminum or silicon-carbide/aluminum composites may be stronger than the interface between the SCS_6 fibers and the Ti-15-3 matrix material. The actual strengths of these interfaces are not well defined at this time and, therefore, can not be compared. However, it can be shown that the titanium matrix composite demands much more strength from the fiber/matrix interface than does an aluminum matrix composite. Analysis of the matrix stress in the loading (0°) direction in the 90° ply of a $[0/90]_{2s}$ laminate is shown in Figure 7. This stress is approximately equal to the stress at the fiber/matrix interface. The aluminum matrix cannot support a load much above its yield strength of approximately 140 MPa. However, the titanium has a very high yield strength, thus much higher loads are carried in the 90° plies and transferred into the fibers. For a given laminate stress level, it is clear that in titanium matrix composites, the fiber/matrix interface is more highly stressed than in aluminum matrix composites. Whereas the three other potential failure modes could be discussed in terms of relative strains to failure of the constituents, the interface failure occurs if the interfacial strength is less then the strength of the matrix and the

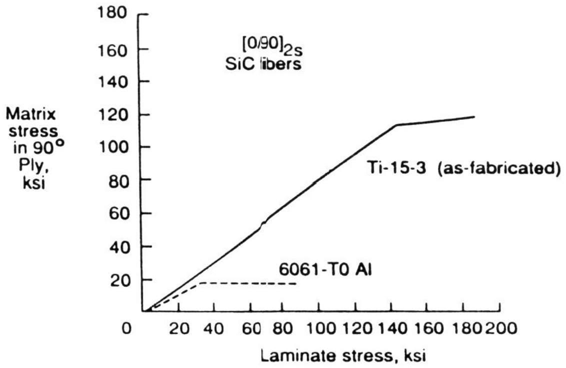

Fig. 7. Predicted matrix stress in 90° plies.

transverse strength of the fiber. In order to realize the full potential of titanium matrix composites, the fiber/matrix interface must be significantly stronger than in the system reported on herein.

SUMMARY

Numerous examples of how fatigue damage can initiate and grow in various MMCs were presented. Based on the relative strain to fatigue failure of the fiber and matrix, the possible damage mechanisms of an MMC can be grouped into three categories: (1) matrix dominated, (2) fiber dominated, and (3) self-similar damage growth. A fourth type of damage development, fiber/matrix interface failure, is dependent on the relative strength of the fiber/matrix interface and the matrix yield strength. These four types of damage were discussed and illustrated by examples. Each damage mechanism may effect the overall stiffness of the composites differently as a function of number of applied load cycles. In most applications of MMC, retention of stiffness is required. Therefore, it is necessary to understand where damage will first initiate, what mechanical properties will be affected, and how to avoid such damage. Clearly, those laminates that contain 0° plies will fail only when the 0° plies fail, thus failure is always fiber dominated. However, as discussed, the initial damage may not be fiber dominated, and as such may not lead directly to laminate failure, but it may cause undesirable losses of material stiffness.

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