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FATIGUE OF CONTINUOUS FIBER REINFORCED METALLIC MATERIALS

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FATIGUE OF CONTINUOUS FIBER REINFORCED METALLIC MATERIALS

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The complex damage mechanisms that occur in fiber reinforced advanced metallic materials are discussed. As examples, results for several lay-ups of SCS-6/Ti-15-3 composites are presented. Fatigue tests were conducted and analyzed for both notched and unnotched specimens at room and elevated temperatures. Test conditions included isothermal, non-isothermal and simulated mission profile thermomechanical fatigue. Test results indicated that the stress in the 0° fibers is the controlling factor in fatigue life for a given test condition. An effective strain approach is presented for predicting crack initiation at notches. Fiber bridging models were applied to crack growth behavior.

INTRODUCTION

Advanced materials are needed to meet the high temperature, low weight requirements established for advanced aircraft propulsion systems and high speed airframes. One class of advanced material is continuous fiber reinforced titanium matrix composites (TMC's). These materials are currently being considered as structural materials for high temperature applications where weight saving is a premium. Stresses are induced in the composite constituents due to temperature change because of the coefficient of thermal expansion mismatch between the fiber and matrix materials. This, coupled with the different strengths and failure modes of the fiber, matrix and fiber/matrix interface, contributes to a very complex problem in predicting and tracking damage initiation and progression in TMC's. Structures built of these materials must satisfy durability and damage tolerance requirements just as any other man-rated aircraft structure. Further, these high temperature structural applications will require that the response of the material to thermomechanical loading be well understood and predictable. This paper will discuss the durability and damage tolerance problems associated with predicting crack initiation and damage growth in notched specimens. In addition, examples of predicted and measured unnotched material response under in-phase and out-of-phase thermomechanical fatigue and simulated mission profile testing will be presented and discussed.

MATERIALS AND TESTING TECHNIQUES

Ti-15-3, a shortened designation for Ti-15V-3Cr-3Al-3Sn, is a metastable beta strip alloy used where cold formability and high strength are desired (1). The composite laminates were made by hot-pressing Ti-15-3 foils between unidirectional tapes of silicon-carbide fibers. These fibers are designated SCS-6 by Textron Specialty Materials, the producer. The fiber

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diameter is 0.142 mm and the fiber is assumed to have isotropic properties with a modulus of 400 GPa. Panels, fabricated by Textron Specialty Materials, of each of the following lay-ups were tested: $[0]_8$, $[0_2/\pm45]_s$, $[0/90]_{2s}$, [0/90/0] and $[0/\pm45/90]_s$. The fiber volume fractions ranged from 32.5 to 39%. A panel of "fiberless" composite was made identically to the composite but without placing the fiber mats between the foils during fabrication. This resulted in a panel of the matrix material that had seen the same thermal and mechanical history during fabrication as the matrix in the composite.

The tests were conducted at NASA Langley Research Center, unless otherwise noted, on an 89 kN servohydraulic test stand. For all composite tests load control was used with a loading rate of approximately 0.89 kN/second for the quasi-static tests and a cyclic frequency of 10 Hz and an R=0.1 for the fatigue tests. Static and fatigue tests of the matrix material were conducted in strain control. Tests were also conducted to determine the rate- and temperature-dependent properties of the matrix material. These tests were conducted using several loading rates and temperatures and under both strain and load control. For room temperature tests, an extensometer with a 25.4-mm gage length was attached to the edge of the specimens to record strain. For the elevated temperature tests, a water cooled, quartz rod extensometer with a 25.4-mm gage length was used. The specimens were heated using a 5kW induction heater with copper coils. During thermomechanical fatigue testing the induction heater and gaseous nitrogen were used in unison to achieve the required cooling rates. The temperature distribution over the gage length was controlled to within $\pm10^oC$ over a 50 mm gage section.

FATIGUE OF UNNOTCHED SPECIMENS

Fatigue damage in unnotched TMC composites was found to be quite dispersed (no single dominant crack) so the overall stiffness and strength of the bulk material were degraded with cycling. The presented work first describes fatigue damage mechanisms at room temperature and at elevated temperature. Next, work analyzing in-phase and out-of-phase thermomechanical fatigue is reviewed. Lastly, material response to a generic hypersonic loading profile is discussed.

Fatigue at Room Temperature

S-N data was experimentally determined at room temperature (RT) for four different lay-ups containing 0^o plies (2). The stress-strain response was monitored during the fatigue life. The laminates containing off-axis plies lost stiffness very early in the cycling history due to the fiber/matrix interface failure. 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 0^o fiber. Since the laminate will not fail until the 0^o fiber fails, it is reasonable to assume that the stress in the 0^o fiber will dictate fatigue life. A plot of the 0^o fiber cyclic stress range against life for the four laminates is shown in Figure 1. The fatigue data from the four different lay-ups were correlated well by the 0^o fiber stress range.

Also plotted in Figure 1 is the strain-controlled fatigue life of the Ti-15-3 matrix material (i.e., the "fiberless" composite). Comparing the fatigue life in terms of the strain range of the 0^o fibers to that of the matrix, it appears that for the same cyclic strain range the fibers would fail first. This was confirmed by examining failure surfaces of fatigued laminates. There was no evidence of significant matrix cracking except in one $[0]_8$ specimen that was cycled at an applied laminate stress of 690 MPa and broke at 516,000 cycles (2). This test is shown as a run-out data point in Figure 1. This specimen had such a long life that the matrix curve may well have intersected the fiber failure curve. Thus, knowledge of the fatigue behavior of the constituents is useful for the prediction and interpretation of the composite fatigue response. The constituent fatigue behavior will provide insight into whether the matrix or the fiber fail first for a given applied cyclic stress. This is especially true for 0^o laminates that are not complicated by fiber/matrix debonding in the off-axis plies.

Fatigue at High Temperature

Unnotched fatigue tests were conducted at 650^oC on the four SCS-6/Ti-15-3 laminates tested at room temperature and the results are reported in detail by Pollock and Johnson (3). Figure 2 compares the maximum strain versus cycles to failure of a unidirectional composite to the maximum strain versus cycles to failure of the Ti-15-3 matrix material loaded in strain control (3). The fatigue test of the unidirectional specimen was considered to be an in-situ fatigue test of the fibers. For high maximum strains (short lives) the initial damage developed in the fibers and the composite had a shorter life than the matrix alone. For low maximum strains (long lives) the initial damage developed in the matrix. Where the two curves intersect, both the matrix and the fibers were equally likely to develop fatigue damage. Figure 3 shows three polished surfaces taken at locations away from the specimen fracture surface for unidirectional specimens with short, medium and long lives. The high stress, short life specimen exhibits many fiber breaks but no significant matrix cracking (the lines visible in the matrix are grain boundaries). The medium life specimen has both fiber breaks and some short matrix cracks. The low stress, long life specimen exhibits no fiber breaks but several long matrix cracks. Therefore, it is reasonable to use the fatigue response of the matrix material and the in-situ 0^o fiber to predict the strain levels at which fiber failure will precede matrix cracking and at which matrix cracking will precede fiber failure in the composite.

Thermomechanical Fatigue

Both thermal and mechanical stresses are developed in the constituents of a laminate during thermomechanical fatigue (TMF). Since constituent stresses are not necessarily related to the measurable laminate strains, an analytical tool is needed to calculate the constituent and laminate behavior of arbitrary lay-ups subjected to arbitrary combinations of mechanical and thermal loading. The *VISCOPLY* program, developed by NASA Langley, is a micromechanics analysis (4) based on constituent properties. The program uses the vanishing fiber diameter (VFD) constitutive model (5) to calculate the orthotropic properties of the plies.

The *VISCOPLY* program was used to predict the laminate response of unidirectional SCS-6/Ti-15-3 composites (4). Neat Ti-15-3 matrix material was tested to determine the required thermo-viscoplastic material constants. The fatigue behavior as a function of maximum applied stress for in-phase and out-of-phase TMF is shown in Figure 4 These tests were conducted at NASA Lewis. The in-phase loadings produced earlier failures compared to the out-of-phase but the specimens lost very little stiffness prior to failure. On the other hand, the out-of-phase loadings resulted in significant stiffness losses due to matrix cracking prior to failure. The *VISCOPLY* program was used to predict the fiber and matrix stresses during the in-phase and an out-of-phase TMF cycle between 93-538^oC and 45-896 MPa (6). The fiber stresses are predicted to be highest for the in-phase test, explaining the earlier laminate fatigue failures. Additionally, the predicted matrix stresses were higher for the out-of-phase loadings, thus explaining the earlier matrix cracking and the resulting stiffness loss measured in ref. 6 during the out-of-phase loadings. When each test shown in Figure 4 was analyzed and the 0^0 fiber stress range was plotted versus the number of cycles to failure, the in-phase and out-of-phase data collapsed into the narrow band labeled TMF in Figure 5.

Other TMF and isothermal data were analyzed and are plotted in Figure 5 (4). Within a given test condition, i.e., temperature, loading frequency, time at temperature, etc., the 0^o fiber stress range seems to correlate with the number of cycles to failure. However, as the test conditions change the fatigue behavior of the 0^o fiber appears to change. Since Figure 5 shows only the stress range in the fiber, the increased loading of the fiber due to matrix stress relaxation is not accounted for. Higher temperatures and slower cycling would both contribute to more load being shifted to the fiber from the matrix. Additional time at temperature could also cause additional fiber/matrix interface reactions that could effect the mechanical behavior (7,8).

Generic Hypersonic Mission Profiles

Mirdamadi and Johnson (9) studied the material response of SCS-6/Ti-15-3 $[0/90]_{2s}$ laminates subjected to a generic hypersonic mission profile. The profile is shown as the insert in Figure 6. The maximum applied temperature was 600°C and maximum applied load was 420 MPa. The stress-strain response of the fifth cycle of the mission profile is shown in Figure 6. The *VISCOPLY* predictions are also shown in the figure. There is excellent agreement between the predicted and measured stress strain response. The prediction was modified empirically to account for the fiber/matrix interface failure and the corresponding reduction in the laminate modulus. The modification procedure, based on isothermal tests, is described in ref. 9. The fiber and matrix stresses under the mission profile loading can now be predicted and related to the constituent fatigue response. An example of the 0^o fiber stress as a function of time is shown in Figure 7. If the total strain range of the laminate measured during the mission profile cycling (approx. 0.0057 from Figure 6) was multiplied by the average fiber modulus over the temperature range in question (approx. 385 GPa), a fiber stress range of 2195 MPa would be calculated. This approach was used to determine the fiber stress range at room temperature in Figure 1. However, the fiber stress range shown in Figure 7 predicted using *VISCOPLY* is approximately 1590 MPa, much lower than that calculated from the overall strain range. This clearly illustrates that the stress in the constituents under TMF loading conditions cannot be directly measured, but must be calculated analytically.

The physical basis for predicting the fatigue life of TMC's under complicated TMF loading is still not fully developed. Perhaps the total stress range in the 0^o fiber is the answer as discussed earlier for simpler loading cases. Or perhaps a linear life fraction model may be applicable (a Miner's Rule approach based on both fiber and matrix damage) as suggested by Russ, Nicholas and Mall (10).

FATIGUE OF NOTCHED SPECIMENS

Damage initiation and growth near local stress concentrations in TMC's is a complex process (11). Under conditions where fibers do not break, damage consists primarily of matrix cracking and fiber/matrix debonding. This phenomena is referred to as fiber bridging. Although the fibers are intact, matrix cracking and fiber/matrix debonding significantly reduce both the stiffness and strength in TMC's (12). Thus both damage initiation and growth are vital issues that must be addressed and understood in order to apply an appropriate design philosophy. In the following sections, a methodology to predict matrix crack initiation is outlined and a fiber bridging model characterization of matrix crack growth is reviewed.

Fatigue Crack Initiation

Hillberry and Johnson (13) developed a methodology to predict the initiation of matrix cracking in notched TMC's based on the Smith-Watson-Topper effective strain parameter (14). Hillberry and Johnson modified the Smith-Watson-Topper effective strain parameter to predict cycles to fatigue damage initiation in the matrix material next to a notch. The modification incorporated the calculated thermal residual stress in the matrix, σ^r, and the orthotropic stress concentration factor, K_t, as follows

$$\Delta\varepsilon_{eff} = [(K_t\varepsilon_{max} + \sigma^r/E_m)\, K_t(\Delta\varepsilon/2)]^{1/2}$$

where E_m is the matrix modulus, ε_{max} is the maximum applied strain and $\Delta\varepsilon$ is the applied strain range.

To illustrate the importance of including the thermal residual stress term, the effective strain parameter was calculated using the above equation with and without the residual stress term for the data from ref. (12 and 13). The results are shown in Figure 8 as a function of the number of cycles to observed crack initiation. The strain controlled matrix fatigue data shown in Figure 1 was replotted in Figure 8 in terms of $\Delta\varepsilon_{eff}$. Better agreement was found between the effective strain parameter for the composite using the thermal residual stress term 0and the matrix material baseline.

Fatigue Crack Growth from Notches

Bakuckas and Johnson (15) conducted analytical and experimental investigations of the effect of fiber bridging on crack growth in center slit $[0]_8$ SCS-6/Ti-15-3 specimens. Under constant amplitude loading the crack growth rate was found to decrease as the crack length increased. The effect of fiber bridging reduced the crack driving force as the crack grew. Since the crack in the composite is growing only in the matrix material, the crack growth rate in the composite should correlate with the crack growth rate in the matrix material alone if the crack driving force in the matrix, ΔK_{mat}, is properly defined. Figure 9 shows crack growth rate versus ΔK data for the Ti-15-3 material. The figure also shows the composite data plotted without accounting for the fiber bridging. The ΔK_{app} does not collapse the composite crack growth data to the Ti-15-3 data showing the need for a definition of ΔK_{mat} that includes fiber bridging.

Several fiber bridging models which combine a continuum fracture mechanics analysis and a micromechanics analysis were investigated (16-18). In all of these models, the intact fibers in the wake of the matrix crack are modeled using a continuous closure pressure. Fiber/matrix debonding is assumed to occur as the crack progresses past each fiber. An unknown constant shear stress τ is assumed to act on the debonded fiber/matrix interface. The fiber bridging models use τ as a fitting parameter. The model proposed by McMeeking (18) provided the most accurate predictions of the measured fiber/matrix debond length and crack opening displacements for test data available, and was used to plot the da/dN data as shown in Figure 9. The values of τ used to fit the data are also shown in Figure 9. Additionally, the values of τ required to fit the da/dN data varied with applied load and crack length. The value of τ required to fit crack opening displacement data or fiber/matrix debond length data were significantly different. If τ were truly a material constant and all the mechanics were properly modeled, τ should not vary for a given material system. Further work needs to be conducted to truly develop a predictive crack growth model for TMC's. In particular, under TMF loading conditions, time- and temperature-dependent material responses will need to be included in the modeling effort.

SUMMARY

This paper summarizes research conducted by the authors and their colleagues on notched and unnotched fatigue of composite laminates made of a titanium alloy matrix reinforced with silicon-carbide fibers (SCS-6/Ti-15-3). The subject research has established a good fundamental understanding of fatigue damage initiation and propagation in continuous fiber reinforced titanium matrix composites at both room and elevated temperatures. The causes of initial damage on both the global and local levels are becoming well defined. Seemingly insignificant factors, such as thermal residual stresses and interfacial strengths, play profound roles in almost every aspect of the fatigue life, from initiation to fracture, and, thus, they must not be overlooked. The time, temperature, material aging, and fiber/matrix interaction effects are complex, and, in some cases, synergistic. This complex interaction is most evident under thermomechanical fatigue, especially with complex mission simulations. Progress is being made in analyzing such materials and loading condition as exemplified by the *VISCOPLY* program. However there is a significant amount of work remaining before a damage tolerance prediction methodology will be available for advanced TMC materials.

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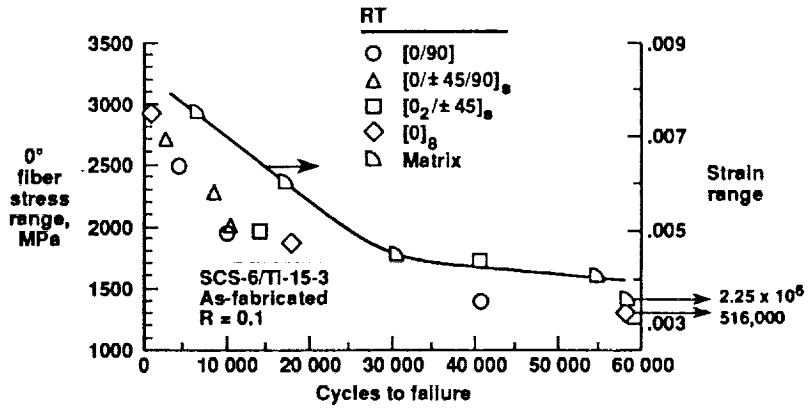

Figure 1. Cyclic stress range in 0° fiber versus number of cycles to laminate failure[2]. Matrix fatigue life is shown in terms of strain range.

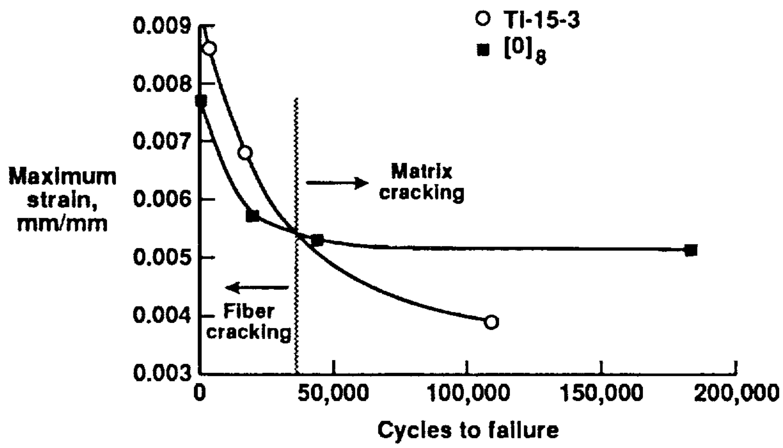

Figure 2. The life of the unidirectional composite and the matrix as a function of maximum strain at 650°C [3].

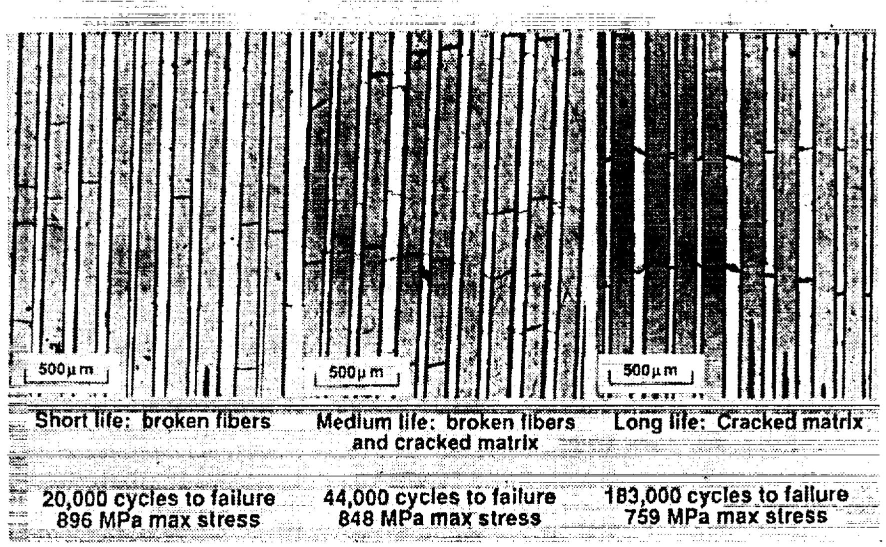

Figure 3. Sections parallel to the loading axis in unidirectional coupons behind the failed surface [3].

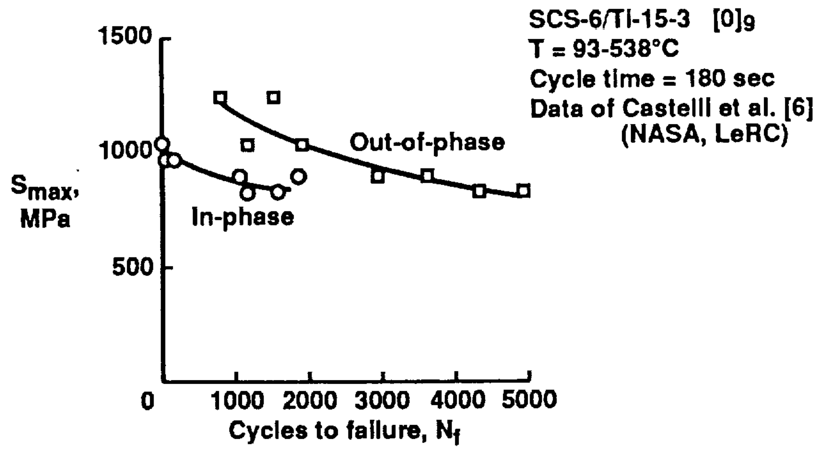

Figure 4. The maximum applied stress as a function of cycles to failure under TMF [(4)].

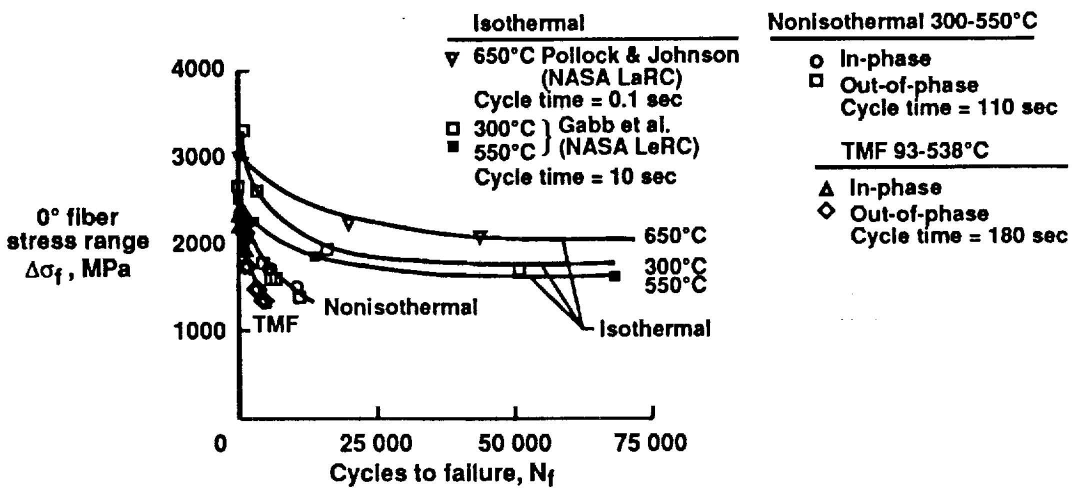

Figure 5. The stress range in the 0° fiber as a function of cycles to failure [(4)].

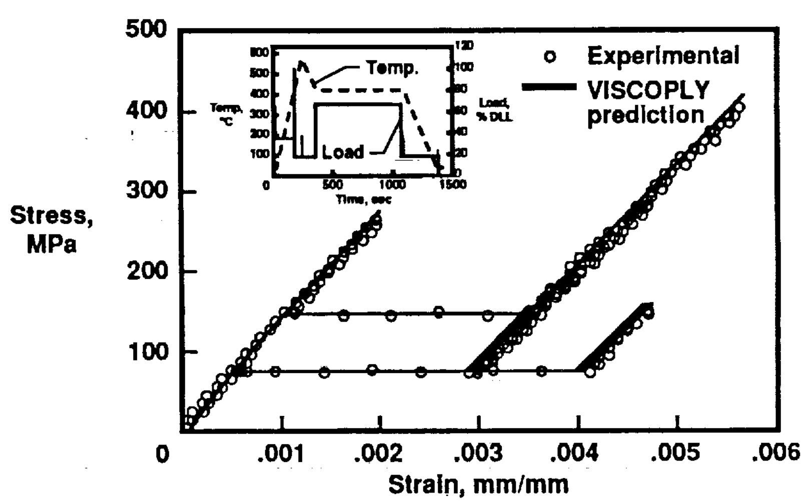

Figure 6. Hypersonic flight mission profile and material measured and predicted stress-strain response [(9)].

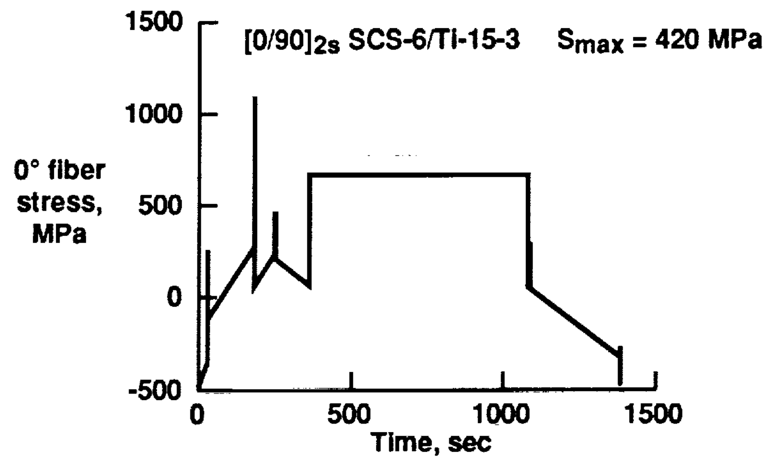

Figure 7. VISCOPLY prediction of 0° fiber stress under flight profile at maximum applied stress of 420 MPa simulating fiber-matrix interface failure of 90° plies [9].

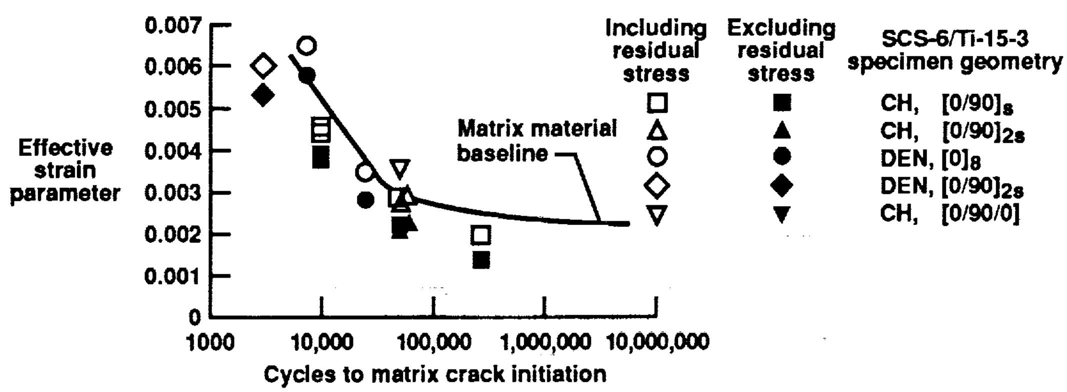

Figure 8. Predicted versus experimental crack initiation [12].

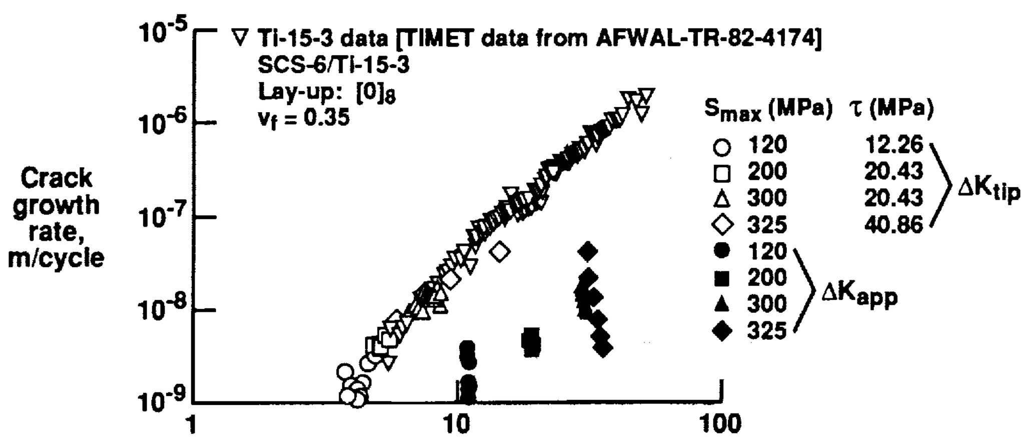

Figure 9. Crack growth rate data from the Ti-15-3 material compared with the crack growth data from the composite. The solid symbols are the composite crack growth data assuming no fiber bridging. The open symbol data are corrected for bridging, assuming the indicated shear stresses [15].

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