Isothermal and Thermomechanical Fatigue of Cross-Ply SCS-6/TIMETAL®21S

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ABSTRACT

The response of a $[0/90]_{2S}$, SCS-6/TIMETAL®21S metal matrix composite under isothermal and thermomechanical fatigue (TMF) environments was investigated. Four sets of tests were conducted involving in-phase and out-of-phase TMF from 150° to 650°C along with isothermal fatigue at both 150°C and 650°C. Experimental results indicated in-phase loading produced shorter fatigue lives than those under out-of-phase loading at high applied stress levels. Conversely, out-of-phase loading produced shorter fatigue lives at lower stress levels. Isothermal fatigue lives at 650°C were longer than those under in-phase conditions by approximately the same factor for all applied maximum stresses. 150°C isothermal loading conditions produced fatigue lives greater than those obtained in the other three cases. Fracture surface examination to determine the dominant failure mechanisms showed that the 150°C and 650°C isothermal fatigue and the out-of-phase TMF all exhibited evidence of matrix cracking, the worst case being the out-of-phase TMF. There was no indication of matrix cracking for the in-phase TMF tests indicating fiber dominant fatigue failure. Examination of strain history and modulus history as well as observations of the very different trends in the fatigue-life curves suggest that different failure mechanisms dominate the out-of-phase TMF behavior in contrast to in-phase and isothermal fatigue behavior.

Analysis, using a linear life fraction model, further suggests that the fatigue lives under the four conditions examined here are controlled by these separate mechanisms.

INTRODUCTION

The need for materials with high strength and high stiffness at high temperatures has in recent years stimulated the development of fiber reinforced metal matrix composites (MMC's). In particular titanium matrix composites (TMC's) have appeared attractive based on specific strength and stiffness /1,2/. However, in envisioned applications, TMC components will be subjected to complex load and temperature profiles. Such conditions, commonly referred to as thermomechanical fatigue (TMF), are expected to be critical for MMC's /3/, and especially for TMC's, where mere thermal cycling can lead to significant, cyclic, constituent stresses and progressive damage due primarily to the coefficient of thermal expansion (CTE) mismatch between the matrix and fiber /4/. Simple TMF loading conditions, which can be produced in the laboratory and can be used as a building block for more complex scenarios, are: (1) in-phase TMF (IP TMF) where maximum-minimum temperature occurs at the time of maximum-minimum load, and (2) out-of-phase TMF (OP TMF) where the load cycle lags the temperature

cycle by half the period, i.e. maximum load occurs at the time of minimum temperature. In recent years, several investigations involving the use of these and other TMF cycles on TMC's have been reported and have demonstrated that thermal and mechanical cycling potentially produce the most damaging fatigue conditions for TMC's /5-9/.

Currently there are several TMC systems under investigation for hypersonic and propulsion applications. Titanium matrix alloys of interest include, but are not limited to, titanium aluminides, $\alpha_2 + \beta +$ orthorhombic, $\alpha + \beta$, and metastable β alloys, all composited with SiC fibers. The SiC/Ti-15Mo-2.6Nb-3Al-0.2Si (at%), commonly referred to as SCS-6/TIMETAL®21S (formerly designated SCS-6/ β 21S) represents a metastable β alloy system developed with a balance between processability and environmental resistance at elevated temperatures.

The objective of this study was to investigate the fatigue behavior under isothermal and thermomechanical fatigue for an SCS-6/TIMETAL®21S cross ply ([0/90]_{2S}) laminate. The applicability of a linear life fraction model, which addresses the separate roles of the fibers and the matrix in the fatigue process, was also examined. Four series of tests were conducted: in-phase TMF, out-of-phase TMF, 650°C and 150°C low frequency isothermal fatigue.

EXPERIMENTAL PROCEDURES

The TMC used for this investigation was a cross-ply laminate, [0.90]_{2S} with an average fiber volume fraction of 0.34. The matrix material was a Ti-15Mo-3Al-2.7Nb-0.2Si (at%) which is a metastable beta titanium alloy, designated as TIMETAL®21S (formerly β 21S). The fiber reinforcement is a silicon carbide (SiC) monofilament, designated SCS-6. This 142 μ m diameter fiber contains a 40 μ m carbon core with a chemically vapor deposited bulk SiC and a double pass carbon rich outer coating designed for use with titanium-based matrices /10/. The foil-fiber-foil technique was used in fabricating the composite panel tested. Foil-fiber-foil refers to the layering of matrix foils, nominally 0.11 mm thick, and fiber mats, nominally 50 fibers per cm, in the appropriate

orientation and consolidating the panel by hot isostatic pressing. A molybdenum cross weave material was employed in the fiber mats in order to maintain adequate fiber spacing and alignment throughout fabrication. Straight sided specimens, nominally 108 mm long x 12.7 mm wide x 1.8 mm thick, were machined using an abrasive water jet cutting technique. Prior to fatigue testing, the edges of the specimens were polished successively on 70 μ m, 45 μ m and 20 μ m diamond wheels. The specimens were then heat treated in vacuum at 621°C for eight hours in an attempt to stabilize alpha precipitation in the matrix.

Testing was performed on two specially designed TMF test systems; one equipped with a servohydraulic cylinder and the other with a pneumatic cylinder for load control. The unique capabilities of the test system are described by Hartman and Russ /11/ and Hartman *et al.* /12/. The test system features a rigid horizontal design, hydraulic friction grips, radiant energy heat lamps, forced air cooling, four zone temperature control, and high temperature extensometry.

In this study four types of tests were conducted: (1) in-phase TMF, (2) out-of-phase TMF, (3), isothermal fatigue at 650°C, and (4) isothermal fatigue at 150°C. All tests were conducted in load control at a frequency of 0.00556 Hz (3 minute cycle) with a triangular waveform and a stress ratio of 0.1. For the TMF tests the temperature was cycled between 150 and 650°C with a constant heating/cooling rate equivalent to the 3 minute cycle (approximately 5.6°C/s). All tests were performed in laboratory air. Periodically throughout each test data acquisition cycles (DAC) consisting of stress, strain, time, and temperature were recorded. These data were used to investigate modulus and strain response as a function of applied cycles. Cycles-to-failure was recorded for each specimen, with failure defined as complete fracture of the sample.

RESULTS AND DISCUSSION

Static Loading

Figure 1 shows the stress-strain response at the maximum (650°C) and minimum (150°C) temperature of the TMF cycles. These tests were run to aid in selection of the stress levels for the fatigue tests. In

addition, the ultimate stresses from these tests provided data for a single cycle-to-failure, where the ultimate stress at the maximum temperature corresponds to an IP or 650°C isothermal fatigue single cycle-to-failure, and the ultimate stress at the minimum temperature corresponds to an OP or 150°C isothermal fatigue single cycle-to-failure. A study by Lerch and Melis /13/ on a similar MMC cross-ply laminate demonstrated that the stress-strain curve from a room temperature static test exhibited two distinct linear regions prior to a third nonlinear region leading to failure. The onset of nonlinearity has been attributed to interface separation at off-axis fibers and matrix plasticity /14-16/. The stress levels at which the nonlinear response occurred were approximately 340 MPa and 380 MPa for 650°C and 150°C, respectively (Fig. 1). In a previous study by Mall and Schubbe /7/, it was noted that the onset of nonlinearity in the elevated temperature static test coincided with the stress at which the IP and OP TMF S-N curves crossed. Therefore, TMF tests in this study were performed at stress levels both above and below this point of nonlinearity on the 650°C static stress-strain curve (340 MPa).

Fatigue

Figure 2 represents the relationship between applied stress and fatigue life (cycles-to-failure) for all four testing conditions. The data are also presented in Table 1. The IP and OP TMF test results indicate similar

Fig. 1: Stress-strain response of [0/90]₂S composite at 150 and 650°C.

Fig. 2: Fatigue life as a function of maximum applied stress.

trends as observed for the SCS-6/Ti-15-3 [0/90]₂S TMC by Mall and Schubbe /7/. The S-N curves from IP and OP TMF tests cross at a stress level coincident to the onset of nonlinearity of the elevated temperature static stress-strain curve. The present composite system, SCS-6/TIMETAL®21 [0/90]₂S, demonstrated longer lives for OP TMF than IP TMF at high stress levels (or above the cross-over point). Conversely, at lower stress levels (or below the cross-over point), IP TMF specimens possessed fatigue lives greater than their OP TMF counterparts. This behavior in MMC's can be attributed to the micromechanical stresses existing in the fiber and matrix as a result of the mismatches in CTE and modulus between matrix and fiber. This mismatch causes the fiber to experience higher stresses (both range and maximum value) under IP TMF in comparison to the OP TMF. Conversely, matrix stresses are more severe under OP TMF conditions. The 650°C isothermal fatigue curve has a similar trend as the IP TMF curve. However, the 650°C isothermal curve is offset by approximately a factor of 10, indicating less severe conditions for the same applied stresses. There were only two 150°C isothermal fatigue tests performed. This was due to the long lives at high stress levels and the limitations of the test capability/specimen design to perform cyclic tests at stress levels greater than 825 MPa. Taking into consideration the limited data, the curve shows a similarity to the OP TMF curve; but, unlike the above comparison, the offset between the 150°C and the OP

curve is much more pronounced, indicating the OP loading condition is a much more extreme environment.

The fatigue life (S-N) data, as a function of maximum applied stress (Fig. 2), can be interpreted with the aid of Fig. 3, which is a schematic of the stresses in the matrix and the 0° fibers. Fiber and matrix stresses are depicted for nominally low and high applied stresses, treating the [0/90]_{2S} composite as a two bar system of 0° fibers and matrix material. The contribution of the 90° fibers is ignored for simplicity and assumed to have the characteristics of the pure matrix. For isothermal fatigue at 650°C, the fiber and matrix stresses increase with higher applied stress, the fiber stresses being larger than those of the matrix because of the fiber's higher modulus. Under IP TMF, the matrix stress range and mean stress remains low, while fiber stress range increases due to the combined mechanical and thermal effect. Comparing then, 650°C isothermal and in-phase TMF data in Fig. 2, the two sets of data are somewhat parallel, but the in-phase lives are considerable shorter for the same applied stress. Since maximum stresses are identical in both cases for a given applied stress, assuming no residual thermal stresses at the high temperature, then the stress range in the fiber appears to be a primary governing factor in determining fatigue life in these two cases. Isothermal fatigue at 150°C has low fiber mean stresses, medium fiber stress range, high mean matrix stresses and medium matrix stress range. For OP TMF, on the other hand, fiber stress range and mean stress are low, while the matrix mean stress and stress range

are much higher and increase with greater applied load. Comparing then, 150°C isothermal and out-of-phase TMF data in Fig. 2, OP TMF lives are considerably shorter for the same applied stress. Since maximum stresses in the matrix are identical in both cases and the mean stresses are higher in the isothermal condition, it would appear that the stress range of the matrix is the governing factor in the fatigue lives for these test environments.

If one compares in-phase and out-of-phase TMF lives, Fig. 3 shows that for high applied stresses the fiber stress range in the in-phase case dominates over that in the out-of-phase case, so that in-phase lives are expected to be lower, as seen in Fig. 2, and dominated by fiber failure. For low applied stresses, the situation is reversed and the matrix stress range is able to dominate for the out-of-phase case, leading to the observed cross-over between the in-phase and out-of-phase data.

Fractography

The examination of the in-phase TMF specimens revealed fracture surfaces characteristic of fiber dominated fracture. The samples showed almost no indication of matrix cracking. One exception was the specimen tested at 280 MPa and lasting more than 16,000 cycles, where small matrix cracks were observed originating at the 90° fibers and extending toward, but not reaching, the 0° fibers. In general the fracture surfaces closely resembled that of a 650°C tension test with dimpling of the matrix, representative of void formation and coalescence, typical of tensile overload. Also, rather extensive fiber pull-out was observed. Figure 4 shows a representative location on the fracture surface of the specimen tested at 400 MPa. The dimpled matrix and pull-out of the 0° fibers (coming out of the page) are clearly discernible in this photograph. These observations, along with the absence of matrix cracking, supports the theory that fracture under the IP TMF environment is dominated by fiber stresses, of which the maximum occurs at the highest temperature, 650°C.

In comparison, evidence of matrix cracking was observed on the fracture surface of the 650°C isothermal fatigue specimen. The specimen tested at 370 MPa, and lasting 6300 cycles was examined, and it

Fig. 3: Schematic representation of micromechanical stresses in fiber and matrix due to low stress followed by high applied stress.

Fig. 4: SEM photograph characteristic of in-phase TMF. Specimen tested at 400 MPa.

was apparent that a crack, or cracks, initiated at or near one cut edge. Matrix cracking was characterized by the relatively flat and featureless appearance of the matrix, typical of transgranular cracking in Ti alloys. The most severely cracked region exhibited matrix cracking through the thickness of the specimen from one surface 0° ply to the other 0° surface ply. This primary crack extended along the width approximately 1 mm, covering less than 8% of the cross-sectional area. Within this region, where matrix cracking encircled 0° fibers, there was little-to-no fiber pull-out. However, fiber pull-out was observed in the overload region.

(a)

(b)

Fig. 6: SEM photographs of specimen tested at 370 MPa, 650°C isothermal fatigue showing 90° ply cracking in the region (a) adjacent to and (b) furthest away from the primary crack.

Fig. 5: SEM photograph showing primary crack for 650°C isothermal fatigue specimen tested at 370 MPa.

Figure 5 shows the extent of the primary crack, which was assumed to initiate at the cut edge. However, for the bulk of the fracture surface only small matrix cracks were observed originating at the 90° fibers and extending toward the 0° fibers. The size of the 90° ply cracks appeared to be a function of the distance from the primary matrix crack, extending just to the 0° fibers, or approximately $60\ \mu\text{m}$, in the region adjacent to the primary crack, Fig. 6a, and only approximately $25\ \mu\text{m}$ at the far edge of the specimen, Fig. 6b. Apparently these smaller cracks grew along the width of the sample (in the direction of the 90° fibers) at a

rate much faster than through the thickness of the specimen (toward the 0° plies). In the regions where matrix cracking did not extend into the 0° plies the matrix exhibited dimpling characteristic of a ductile tensile overload. It should be noted that fiber pull-out was observed in the tensile overload regions, but to a lesser extent than witnessed from IP TMF. These observations suggest both fiber and matrix damage mechanisms are present in the 650°C isothermal test environment.

The OP TMF test specimens exhibited fracture surfaces considerably different from both the IP and 650°C tests. Evidence of void formation in the matrix was not observed. In contrast the appearance of the matrix material was very planar and faceted, indicative of many cracked regions, Fig. 7. The fracture surface suggests that many small cracks initiated and grew throughout the matrix. Another notable difference in the fractured surface was the lack of fiber pull-out. From the flat nature of the fracture surface (fibers failing in the crack plane) one can speculate that fiber bridging may not be significant for OP TMF. The lack of fiber pull-out and the extent of matrix cracking corroborate the hypothesis that fracture under OP TMF is dominated by matrix damage mechanisms. Also, it should be noted that there were no noticeable changes in the appearance of the fracture surfaces as a function of stress level for the OP TMF samples examined.

The 150°C specimen examined showed characteris-

tics similar to both the OP TMF and the 650°C isothermal tests. Similar to OP TMF, significant matrix cracking in the form of many small cracks was observed, Fig. 8a. However, similar to the 650°C test, there were regions where the cracking was restricted to the 90° plies and dimpling of the matrix, characteristic of tensile overload, was present around the 0° fibers, Fig. 8b. However, fiber pull-out was not prominent for the 150°C fatigue sample, and it was observed that the fracture of the fibers appeared more severe as compared to the other three test environments.

Fig. 7: SEM photograph characteristic of out-of-phase TMF. Specimen tested at 300 MPa.

Fig. 8: SEM photographs of specimen tested at 650 MPa, 150°C isothermal fatigue showing matrix cracking (a) throughout 0° plies and (b) limited only to 90° ply.

Stress-Strain Response

Maximum and minimum strains as well as stress-strain behavior were recorded periodically during all tests. The strain presented is the mechanical component of the total strain measured by the extensometer. The total strain consists of both mechanical and thermal components. The thermal component is obtained from the coefficient of thermal expansion of the composite which was measured prior to mechanical cycling and found to be constant over the temperature range of 150°C to 650°C with an average value of $6.5 \times 10^{-10}/^{\circ}\text{C}$.

The history of the maximum and minimum strains during the fatigue tests provides useful information about the MMC's response and damage mechanisms during the test. Figs. 9, 10, 11, and 12 show typical

Fig. 9: Maximum and minimum strain history, in-phase TMF, 340 MPa.

Fig. 10: Maximum and minimum strain history, 650°C isothermal fatigue, 400 MPa.

Fig. 11: Maximum and minimum strain history, out-of-phase TMF, 340 MPa.

Fig. 12: Maximum and minimum strain history, 150°C isothermal fatigue, 750 MPa.

strain behaviors of IP TMF, 650°C isothermal, OP TMF, and 150°C isothermal tests, respectively. The difference between the two curves of maximum and minimum strains is an indication of the stiffness of the material. If the curves separate, as in Fig. 11, it implies damage accumulation in the form of either fiber breakage or matrix cracking, or both. If the maximum and minimum strains increase together as in Figs. 9 and 10, this implies inelastic deformation type damage and appears to indicate accumulated creep strains in the matrix material, assuming the fibers do not creep at these temperatures. In the 650°C isothermal case for all stresses, a small amount of creep occurs over the first few cycles as illustrated in Fig. 10 for an applied stress

of 400 MPa. For several hundred cycles, strain continues to accumulate gradually, which is attributed to creep in the matrix since the stiffness changes are almost negligible. As the strains continue to increase, a greater portion of the load is transferred to the 0° fibers. Eventually, the increased fiber stress causes the initiation and accumulation of fiber breakage leading to ultimate failure. The decrease in stiffness during this process is still rather small, thereby indicating that the number of fiber breaks is probably small until just prior to final failure.

At the lower stress levels of 340 and 370 MPa in the isothermal tests, the accumulation of strain beyond the first few cycles is very slow, indicating that no extensive matrix creep is occurring. Figure 13 shows the results at 370 MPa where the total life is beyond 6000 cycles. Note that the minimum strain is increasing, a point which will be discussed later. After a large number of cycles the creep, or matrix damage, causes the fibers to reach a level where fiber breakage occurs rapidly. The rapid increase in strain at about 6000 cycles in Fig. 13 is an indication of the fiber breakage process which appears to take only a few hundred cycles, therefore for a long life test this is a small fraction of the total life.

Strain accumulation data for the in-phase tests, Fig. 9, appear to indicate a similar pattern as in the isothermal tests. However, the damage accumulation is more rapid for a given applied stress due to the larger fiber stress range caused by the additional thermal strains as explained previously. The creep process should result in increasingly higher maximum stresses in the fibers if there is no fiber failure. This transfer of load from matrix to fiber with accumulated creep should continue at an ever decreasing rate until either the stress range in the fibers reaches a critical level where fiber failure initiates or the creeping phenomenon ceases. At all stress levels in the IP TMF tests, there exists an initial stage over a limited number of cycles where creep in the matrix redistributes the applied stresses to the fibers. Beyond that, creep continues to occur at a slower but relatively constant rate. This continued creep does not appear to be consistent with the idea of matrix stresses decreasing continuously as fiber stresses increase with accumulated strain. For example, at the very low stress of 280 MPa, Fig. 14, the

Fig. 13: Maximum and minimum strain history, 650°C isothermal fatigue, 370 MPa.

Fig. 14: Maximum and minimum strain history, in-phase TMF, 280 MPa.

apparent creep ratchetting continues to occur gradually over several thousands of cycles before accelerating over the last few hundred. This behavior is similar to that observed in the isothermal case at 370 MPa, Fig. 13, where initial creep occurs and creep ratchetting occurs gradually over thousands of cycles. While it can be presumed that environmental degradation of the interface plays some role in the strain accumulation behavior, particularly in the long-term tests, it is also postulated that occasional fiber breakage must occur throughout the test to account for the steady increase in strain. This aspect is discussed further later.

The apparent creep ratchetting during the isothermal and in-phase TMF tests may have been

caused by the matrix experiencing the maximum load at higher temperature, 650°C. This condition is absent in the out-of-phase and 150°C isothermal tests (Figs. 11 and 12) where the maximum load occurs at lower temperature, 150°C. Unlike the IP and 650°C isothermal tests the strain histories of the OP and 150°C isothermal tests show no rapid increase in the initial stage. In contrast, the strains are stable during the initial stages of the fatigue life; as the tests progress the maximum strains increase and the minimum strains are relatively unchanged. This trend is more pronounced in the out-of-phase cases. This is expected in OP TMF tests, since the matrix experiences the highest stress range in these tests as mentioned earlier. In the last stage of these tests, the maximum strain increases at an accelerated rate until eventual failure. At this point the matrix has cracked almost completely, as a result of the matrix stress range, and was observed in the fractured specimen. At this stage, the load is then shed completely to the fibers which, in turn, fail rapidly. It must be noted that the failure locations of the 150°C tests were not captured within the gage length, as defined by the extensometer length. Therefore, the strain measurements of Fig. 12 do not adequately reflect the severity of damage under this test condition.

Modulus History

In addition to using the difference between minimum and maximum strain as a measure of stiffness, modulus was calculated from load-displacement measurements acquired over a complete cycle periodically during the tests. Figures 15, 16, and 17 show the typical variation of Young's modulus as a function of fatigue cycles for the IP TMF, OP TMF, and 650°C isothermal fatigue tests, respectively. The 150°C isothermal fatigue tests were omitted because failures occurred outside the gage section. The moduli were calculated using the initial linear portion of the stress-strain curves. The modulus shows little or no change during the IP tests, Fig. 15, as is also demonstrated in the minimum and maximum strain plots, Figs. 9 and 14. This would seem to indicate that the specimens subjected to IP TMF tests do not experience fatigue damage accumulation. However, if the failure process involves creep ratchetting of the matrix, then continued strain accumulation should not

Fig. 15: Modulus history for in-phase TMF tests.

Fig. 16: Modulus history for out-of-phase TMF tests.

Fig. 17: Modulus history for 650°C isothermal fatigue tests.

occur since the fibers accommodate higher and higher stresses. The small drop in modulus in the two shorter term tests (400 and 340 MPa) shown in Fig. 15, may be an indication of small amounts of fiber breakage, an observation that is not apparent from the maximum and minimum strain plots. Modulus reduction in the higher stress tests does not appear to occur. On the other hand, the failure process under OP TMF is dominated by fatigue of the matrix and damage accumulation as evidenced by a substantial reduction in modulus during a large part of the test, as shown in Fig. 16. The OP tests all show essentially constant modulus during the early phases of the test, followed by an accelerated reduction in modulus prior to failure. The 650°C isothermal fatigue tests produce anomalous behavior as illustrated in Fig. 17. Here, for stress levels below 370 MPa where lives exceed thousands of cycles, the modulus increases after several hundred cycles. This increase in modulus is attributed to a thermal instability of the titanium matrix which results in alpha (hexagonal close packed structure) phase precipitation within the beta (body centered cubic crystal structure) phase. This can occur with the specimens tested under isothermal fatigue at low stresses, because they are subjected to long durations at a temperature of 650°C before any appreciable fatigue damage occurs. Such a trend is not evident in the IP TMF tests because, although comparable lives were achieved in some tests, the specimens are not subjected to the long durations at high temperature due to the thermal cycling. This phenomenon of composite stiffening with long exposure time at high temperatures was also witnessed previously by Mall and Schubbe /7/ in their investigation of the SCS-6/Ti-15-3 composite system.

The apparent discrepancy between the postulated behavior of fiber breakage causing additional strain accumulation with no corresponding reduction in stiffness can be partially explained through the use of a simple model /17/. A unit cylindrical cell with a broken fiber at the center and a region of fiber/matrix cracking or debonding adjacent to the break is modeled in a simple manner using simple strength of materials concepts. If the fiber break is allowed to result in a small separation between the broken fibers, resulting in plastic or creep deformation in the adjacent matrix, then computations show that the ratio of the elongation

of the unit cell with the broken fiber to that with the unbroken fiber can be considerably higher than the corresponding relative change in stiffness. The simple model, validated with finite element computations /17/, provides a plausible explanation of the observations of strain and modulus history in the in-phase TMF tests and also helps to explain the results in the isothermal tests at 650°C. The proposed mechanisms for the continued strain accumulation in the IP TMF tests, with no corresponding decrease in stiffness, is the gradual accumulation of fiber breakage throughout the test. This mechanism has been validated recently with acoustic emission measurements that detect individual fiber breaks during in-phase TMF but observe no such breaks in OP TMF tests /18/.

Fatigue Life Predictions

In an attempt to gain more insight into the modes of failure of the tested TMC laminate under isothermal and TMF conditions, cycles-to-failure were compared to different parameters calculated from the micro-mechanics analysis using the METCAN program /19/. In earlier work /9/ parameters investigated for correlation included applied stress range, maximum applied stress, mechanical strain range, fiber axial stress range, fiber maximum axial stress, matrix maximum effective stress, matrix effective stress range, and matrix effective strain range. No single parameter was found to be a correlating parameter which could consolidate all the data from IP and OP tests on a unidirectional titanium aluminide composite /9/ or this material /20/. Further fractographic observations show that fatigue mechanisms are controlled primarily by the fiber in IP TMF, by the matrix in OP TMF tests, and some combination of the two for the isothermal fatigue tests. Therefore, a linear life fraction model (LLFM) which consists of a term representing a fiber dominated mode and a second term representing a matrix dominated mode is employed. The LLFM developed in /9/ has the form

$$\frac{N}{N_f} + \frac{N}{N_m} = 1 \quad (1)$$

where N_f and N_m are the number of cycles to failure solely due to the fiber dominated mode and the matrix

dominated mode, respectively, and N is total life or number of cycles-to-failure. Similar to the earlier work /9/, N_f and N_m are assumed to be of the following form:

$$N_f = 10^{N_0} \left(1 - \frac{\sigma_{\max}(1-R)^m}{\sigma^*} \right) \quad (2)$$

$$N_m = B (\Delta\sigma_m)^{-n} \quad (3)$$

where $\sigma_{\max}$ is the maximum stress in the 0° fibers, R is the stress ratio in the fiber, σ_m is the average axial stress range in the matrix, σ^* is the maximum fiber stress at failure in a tension test as calculated from the METCAN program /19/, and N_0 , m , B and n are empirical constants. In /20/, it became apparent that neither maximum stress in the fiber nor stress range in the fiber is an ideal correlating parameter, particularly when considering isothermal as well as TMF data. Therefore, an expression for stress in the fiber that accounts for the combined influence of maximum stress and stress range is introduced in eq. (2) through the term $\sigma_{\max} (1-R)^m$, where m is an empirical constant between 0 and 1. This implies that fatigue life is governed by some combination of stress range and level of mean stress. To account for the different failure stresses under static loading at 150°C and 650°C , σ^* takes on a different value at each temperature depending on the temperature at which the maximum stress occurs in the cycle. The empirical constants (N_0 , B , n , and m) are obtained by trial and error using the test data. The experimental data from the in-phase and out-of-phase TMF tests and the isothermal data at the two temperatures can be relatively well predicted from this model using constants similar to those in /20/. The values obtained in this study with the aid of an error minimization routine using the test data are: $N_0 = 9.0$, $B = 2.0 \times 10^{23}$, $n = 7.8$, $m = 0.44$ and $\sigma^* = 1450 \text{ MPa}$ (650°C) or 2400 MPa (150°C).

In all of the life predictions, the stresses in the 0° fibers at maximum and minimum load were determined from the METCAN program /19/. The life predictions for isothermal and TMF data are shown in Fig. 18 where measured fatigue lives are plotted against the predicted values. Perfect correlation would be represented by a straight line as shown in the figure. It can be seen that the model does a reasonably good job for all four test environments. The good correlation was

Fig. 18: Comparison of experimental and analytical predictions of fatigue life.

obtained only after the introduction of different values of σ^* for the 150°C isothermal test. Without this, the low temperature isothermal data would be grossly underpredicted based on constants obtained for the other three data sets in /20/. This shows that the fiber stresses do not correlate fatigue lives, even when the failure mode is fiber dominated. The temperature at which the maximum fiber stresses occur must be accounted for, indicating that environmental effects also play a role in the fatigue process.

The good correlation of predicted and observed fatigue life is based on stresses calculated from the first cycle, even though it is observed that matrix creep, matrix fatigue, damage in the $[90^\circ]$ plies, and progressive fiber failure can all occur during the life of the various test specimens. This observation seems to indicate that once the initial stresses are established, the remaining processes which may involve load transfer between fiber and matrix are fairly well determined.

CONCLUSIONS

A study was conducted to investigate the thermo-mechanical fatigue behavior of a cross-ply laminate of silicon carbide fiber reinforced titanium matrix composite, SCS-6/TIMETAL®21S. Four sets of tests involving in-phase and out-of-phase TMF from 150°C to 650°C along with isothermal fatigue at 150°C and

650°C were performed. This study led to the following conclusions:

1. The fatigue life for in-phase TMF conditions at higher stresses is considerably less than those obtained under the comparable out-of-phase TMF conditions. On the other hand, in the case of low stress in-phase TMF conditions, fatigue life was increased considerably from that obtained under comparable out-of-phase conditions. This caused the cross-over of in-phase and out-of-phase fatigue life curves, and it occurred at a stress level which was equivalent to the onset of nonlinearity on the 650°C static stress-strain curve. The 650°C isothermal fatigue life was greater than those obtained under in-phase TMF conditions and the 150°C isothermal fatigue life was considerably greater than those obtained under out-of-phase TMF conditions, which confirms previous findings that thermomechanical fatigue can produce the most dangerous conditions for some MMC's during their applications.
2. Fracture surface examination to determine the dominant failure mechanisms showed that the 150°C and 650°C isothermal fatigue and the out-of-phase TMF all exhibited evidence of matrix cracking, the worst case being the out-of-phase TMF. In the case of the 650°C isothermal fatigue tests a primary crack was present leading to final fracture of the fibers, and fiber pull-out, in the uncracked regions. The in-phase TMF specimens revealed almost no matrix cracking and large amounts of fiber pull-out, indicating fiber dominant fatigue failure. In contrast, the out-of-phase TMF specimens demonstrated insignificant amounts of fiber pull-out and negligible regions of tensile overload.
3. The analysis of strains monitored during four sets of tests provided the complete description of their fatigue damage mechanisms. The fatigue damage mechanisms during the 650°C isothermal fatigue and in-phase TMF tests were dominated early by creep-ratchetting of matrix followed eventually by fiber fracture. The gradual accumulation of strain during in-phase TMF is attributed to fiber breakage, even though no accompanying decrease in modulus is observed. The 150°C isothermal and out-of-phase

tests appear to be governed by progressive matrix cracking until the fiber stresses reach a critical value at which point they fail rapidly.

REFERENCES

1. J.M. Larsen, W.C. Revelos and M.L. Gambone, *Materials Research Society Symposium Proceedings*, 273, 3-16 (1992).
2. J.M. Larsen, S.M. Russ and J.W. Jones, *AGARD Conference Proceedings*, AGARD Panel on Structures and Materials, Bordeaux, France, September 1993 (in press).
3. G.R. Halford, B.A. Lerch, J.F. Saltsman and V.K. Arya, *Thermomechanical Fatigue Behavior of Materials*, ASTM STP 1186, H. Sehitoglu (ed.), American Society for Testing and Materials, Philadelphia, PA, USA, 1993, pp. 176-194.
4. S.M. Russ, *Met. Trans.*, 21A, 1595-1602 (1990).
5. T.P. Gabb, J.G. Gayda and R.A. MacKay, *J. Comp. Materials*, 24, 667-686 (1990).
6. M.G. Castelli, J.R. Ellis and P.A. Bartolotta, "Thermomechanical Testing Techniques for High Temperature Composites: TMF Behavior of SiC (SCS6)Ti-15-3", NASA Technical Memorandum 103171, NASA Lewis Research Center, 1990.
7. S. Mall and J.J. Schutte, *Comp. Science Tech.*, (in press).
8. B.S. Majumdar and G.M. Newaz, *Composite Materials: Fatigue and Fracture* (Third Volume), ASTM STP 1110, T.K. O'Brien (ed.), American Society for Testing and Materials, Philadelphia, PA, USA, 1991, pp. 732-752.
9. S.M. Russ, T. Nicholas, M. Bates and S. Mall, *Failure Mechanisms in High Temperature Composite Materials*, ASME AD, Vol. 22 and AMD, Vol. 122, 1991, pp. 37-43.
10. G. Das, *Sixth World Conference on Titanium, Part II*, P. Lacombe, R. Tricot and G. Berangue (eds.), Les Ubis Cedex, France: Les Editions de Physique, 1988, pp. 907-912.
11. G.A. Hartman and S.M. Russ, *Metal Matrix Composites: Testing, Analysis and Failure Modes*, ASTM STP 1032, W.S. Johnson (ed.), American Society for Testing and Materials,

- Philadelphia, PA, USA, 1989, pp. 43-53.
12. G.A. Hartman, L.P. Zawada and S.M. Russ, *Proceedings, Fifth Annual Hostile Environments and High Temperature Measurements Conference*, Society for Experimental Mechanics, Bethel, CT, USA, 1988, pp. 31-38.
 13. B.A. Lerch and M.E. Melis, "Experimental and Analytical Analysis of Stress-Strain Behavior in a $[90^{\circ}/0^{\circ}]_{2S}$ SiC/Ti-15-3 Laminate", NASA Technical Memorandum 104470, NASA Lewis Research Center, August 1991.
 14. W.S. Johnson, S.J. Lubowinski and A.L. Highsmith, *thermal and Mechanical Behavior of Metal Matrix and Ceramic Matrix Composites*, ASTM STP 1080, J.M. Kennedy, H.H. Moeller and W.S. Johnson (eds.), American Society for Testing and Materials, Philadelphia, PA, USA, 1990, pp. 193-218.
 15. W.D. Pollock and W.S. Johnson, *Composite Materials: Testing and Design* (Tenth Volume), ASTM STP 1120, Glenn C. Grimes (ed.), American Society for Testing and Materials, Philadelphia, PA, 1992, pp. 175-191.
 16. G.M. Newaz and B.S. Majumdar, *Failure Mechanisms in High Temperature Composite Materials*, ASME AD-Vol. 22 and AMD-Vol. 122, 1991, pp. 55-65.
 17. T. Nicholas and J. Ahmad, "Modeling fiber breakage in a metal matrix composite", *Comp. Sci. Tech.* (in press).
 18. R.W. Neu and I. Roman, "Acoustic emission monitoring of damage in metal matrix composites subjected to thermomechanical fatigue", *Comp. Sci. Tech.* (in press).
 19. C.C. Chamis, P.L.N. Murthy and D.A. Hopkins, *Thermal and Mechanical Behavior of Metal Matrix and Ceramic Matrix Composites*, ASTM STP 1080, J.M. Kennedy, H.H. Moeller and W.S. Johnson (eds.), American Society for Testing and Materials, Philadelphia, PA, USA, 1990, pp. 56-69.
 20. S. Mall, D.G. Hanson, T. Nicholas and S.M. Russ, *Constitutive Behavior of High Temperature Composites*, MD-Vol. 40, B.S. Majumdar, G.M. Newaz and S. Mall (eds.), American Society of Mechanical Engineers, New York, USA, 1992, pp. 91-106.