Thermomechanical Damage Development in SiC(SCS6)/Ti-6-4 Metal Matrix Composite

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A series of experiments were conducted to investigate the thermomechanical fatigue behavior of SiC(SCS-6)/Ti-6-4 composite. Three types of tests were conducted. In the first, specimens were subjected to load-controlled mechanical cycling under isothermal conditions. The other two types of test involved simultaneous cycling of load and temperature: in-phase and out-of-phase cycles. The effect of temperature cyclic ranges of 250°C to 528°C were evaluated for the out-of-phase (low temperature-high stress) conditions. A single temperature range of 250°C was employed for the in-phase (high temperature-high stress) condition. Fatigue-life diagrams were developed to evaluate the fatigue performance of the composite based on certain damage mechanism maps. In addition, micromechanical stresses in the fiber and matrix were computed using a generalized plane strain finite element analysis. The intention of this analytical effort was to provide the understanding of the fundamental mechanisms governing material behavior for guiding the development of life prediction methodology.

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INTRODUCTION

Metal matrix composites are currently being considered for advanced gas turbine components because of their high strength-to-density ratio at high temperatures. For safe application of these new materials, their failure mechanisms and modes must be understood and a life prediction method which incorporates them into the design process must be developed. A sound approach to achieve this objective is to first understand the mechanisms of damage accumulation and to then, with this knowledge, develop models that can form a basis for calculation of the fatigue life. The investigation of damage accumulation mechanisms and failure modes requires a data base to develop the relationship between macromechanical parameters and fatigue initiation resistance. The selection of experimental procedure, which simulates the component operating environment, is the most critical step in the development of a data base.

Hot section components of aircraft gas turbine engines are subjected to severe thermal-structural loading conditions during the engine mission cycle. The most damaging stresses and strains are those induced by the steep thermal gradients, which occur during the start-up and shutdown transients. The thermal mechanical fatigue (TMF) damage produced by this thermal mechanical cycling can often shorten component life. Isothermal fatigue data obtained at various temperatures cannot be readily applied to TMF loading conditions because the mechanisms of fatigue are often different at the extremes of temperature in the TMF cycle. Thermal mechanical fatigue damage can be investigated by independently cycling mechanical load (stress or strain) and temperature, fatigue "in-phase" or "out-of-phase" (1). For this study, a test is defined as in-phase when the maximum temperature coincides with the maximum positive stress. In contrast, a test is defined as out-of-phase when the maximum temperature coincides with the minimum stress and the minimum temperature coincides with the maximum stress. Figure 1 illustrates the temperature and stress waveforms used for an out-of-phase and in-phase tests and their phasing relationship. TMF testing has been used to study fatigue resistance of many monolithic structural alloys (2,3). Several approaches to the prediction of material fatigue behavior under TMF conditions have been developed (2,4). The resulting complexity of the experimental data has precluded the development of a general model of thermal mechanical fatigue. Therefore, it is necessary to consider each MMC system individually and to develop relationships in terms of appropriate parameters, which will allow predictions of structural behavior and life and identification of the important modes and mechanisms of failures.

The past decade has seen significant advancement in understanding the fundamental mechanisms governing material life and identification of the important modes and mechanisms of failures. In Section II of this paper, a brief review of the composite damage mode reported in the literatures is presented. In Section III, the material description and experimental procedures were given. In Section IV, the results of fractographic analysis is discussed. In Section V, an attempt is made to correlate change in compliance with damage accumulation. The maximum and minimum strain were monitored to define damage accumulation mechanisms. In Section VI, the fatigue-life diagrams were constructed to evaluate fatigue performance of a SiC/Ti-6-4 composite system. The second objective was to understand the fundamental mechanisms governing material behavior to guide the development of life prediction methodology and provide more quantitatively accurate results to the investigation. For this purpose, a detailed numerical solution using a generalized plane strain finite element analysis was performed to predict micromechanical stresses in the fiber and matrix.

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Under cyclic loading of composites, the matrix For SiC reinforced titanium alloy composite systems, fiber breakage occurs at applied stresses exceeding 80% of the fiber strength strains for the fibers. This cycling modes of MMC's into three categories: (a) fiber-dominated, (b) matrix-dominated, (c) fiber/matrix interfacial dominated. The following section is intended to give a brief description of failure modes reported in the literatures.

A. Fiber-Dominated Damage

In this mode, fibers will be broken randomly during the early stage of fatigue loading. During cyclic loading, a broken fiber causes shear-stress concentration at the matrix interface. The interface may then fail, leading to debonding of the fibers from the surrounding matrix. The debonded area acts as a stress concentration site for the longitudinal tensile stress. The magnified tensile stress may exceed the fracture stress of the matrix leading to a transverse crack in the matrix. For example, fiber fracture was found to be the dominant damage mechanism controlling the fatigue behavior of FP/A1 (7). For SiC reinforced titanium alloy composite systems, fiber breakage occurs at applied stresses exceeding 80% of the fiber strength in the composite (9).

B. Matrix-Dominated Damage

Under cyclic loading of composites, the matrix is subjected to strain-controlled fatigue due to the constraint provided by the fibers, and it yields at strain levels below the fracture strains for the fibers. This cycling initiates cracks on the surface of matrix. These cracks grow until they strike an interface, if the stress at the crack tip is insufficient to break the fiber. At sufficiently low strains the cracks may be confined to the matrix alone, increasing only in numbers. At higher applied strains, the fibers at the crack tip may fail, allowing the matrix cracks to grow, and finally the composite fails due to tensile overload.

C. Fiber/Matrix Interfacial Dominated Damage

If the applied cyclic stress range is higher than the cracking stress of the interfacial reaction layer, a crack will initiate at this brittle layer. Under cycling loading, the crack will propagate into the matrix and will be bridged by the unbroken fibers. The unbroken fiber will cause fiber failure.

MATERIAL AND EXPERIMENTAL PROCEDURE

The composite system considered in this investigation consists of silicon carbide fibers (SCS-6 produced by Textron Specialty Materials) in a Ti-6Al-4V matrix. It was fabricated in the form of panels fabricated using the foil-fiber-foil method. The lay-up consisted of eight plies of continuous 0° (longitudinal) oriented fibers in a matrix providing a nominal volume fraction of 35 percent. The overall panel thickness was 1.78 mm. Longitudinally oriented test specimens, Figure 2, were cut from the panels using wire EDM. The cut surfaces were then polished to remove any surface damage.

All tests were conducted on a computer controlled, servo-hydraulic test system employing direct induction heating with forced air cooling cycles. All tests were performed on specimens with the fiber oriented in line with the loading axis and in an air environment. All experiments were run to failure under load control, tension-tension conditions with a minimum/maximum stress ratio (R) of 0.05. Here failure was defined as complete fracture (separation of the specimen into two pieces). The test matrix, shown in Table I, was formulated to investigate the behavior of the composite over its full working temperature spectrum. Throughout each test the maximum and minimum strain were monitored in an attempt to correlate change in compliance with damage accumulation.

The effects of temperature range for temperature cyclic range of 250°C to 528°C were evaluated for the out-of-phase (low temperature-high stress) conditions. A single temperature range of 250°C was employed for the in-phase (high temperature-high stress) condition. The initial load or stress was established by obtaining a maximum strain value of 0.50 percent at the peak point of the initial cycle and a minimum strain of 0.05 percent at minimum point of the initial cycle. This yielded a strain range of 0.45 percent at
the start-up point for all the tests involved in the study. One exception to this was an in-phase test, which was started at a strain range of 0.647 percent. The isothermal tests were conducted at 371°C, with a minimum/maximum stress ratio of 0.05.

Heating rates (T) ranging from approximately 4.4°C per second to approximately 14.4°C per second and stress rates (d) ranging from 7 MPa per second to approximately 55 MPa per second were evaluated. Due to the length of time required to remove thermal energy from the test specimen at lower temperature, the test specimen diameter was governed by a fatigue crack growth in the matrix (Figure 4c). This region did not show strong fiber debonding and ductile dimpling of the matrix.

Fractography conducted on isothermal fatigue specimens is shown in Figure 4. The fracture surface at the edge of specimen exhibited relatively flat regions, representative of crack growth in the matrix (Figure 4c). This region did not show strong fiber debonding and pullout, which was characterized by fiber pullout, debonding and ductile dimpling of the matrix.

SEM examination of the fracture surface of the out-of-phase test specimens revealed significant differences in appearance depending on the cycle temperature range. For low AT (250°C), interface failure was followed by fiber fracture and pullout, as seen in Figure 5. The Ti-6-4 matrix showed a dimpled appearance typical of tensile rupture (Figure 5b). Few surface cracks were developed (Figure 5d). SEM examination at the edge of the specimen showed no fiber debonding and pullout in the crack propagation site, as seen in Figure 5c.

An SEM photo of the fracture surfaces of specimens tested at high AT (528°C) are shown in Figure 6. For this temperature range, fatigue crack initiation occurred in the Ti-6-4 matrix and numerous surface initiated cracks were clearly visible, as seen in Figure 6a. The fracture surface showed an oxidized zone at the periphery. In this zone the surface was essentially flat in appearance with the fiber sheared off (Figure 6b). In the center of the specimen, where fatigue occurred, extensive fiber pullout was evident with the matrix exhibiting a dimpled rupture appearance. This was followed by shear of the fiber as the crack approached or met the fibers. On the fracture surface of this fatigue crack was large enough, tensile rupture occurred.

### TEST RESULTS

During each test, the maximum and minimum strain were recorded in an attempt to correlate change in compliance with damage accumulation. The strain history of the out-of-phase cycle, which performed under temperature extremes of 121°C and 454°C, is shown in Figure 7. As seen in this figure, the strain range and mean strain values increased as cycles accumulated. The increase in strain range is hypothesized to be due to the damage generated in the matrix. The strain history of the in-phase cycle with temperature extremes of 121°C and 371°C, is shown in Figure 8. For in-phase tests, the strain range remained relatively unchanged, while the mean strain increased in the early stages of the test. The accumulated strain comparison between in-phase and out-of-phase TMF tests cycled from 121°C to 371°C, is given in Figure 9. A significant decrease in bulk modules values was noted for cycles beyond the crack initiation point. It was also noted that only a small amount of mean strain shift was observed in the early stages of the out-of-phase test, whereas significant ratcheting was seen to occur in the in-phase test. After stabilization at 15,000 cycles, the in-phase strain ratcheting response revealed little change in the original compliance as cycles accumulated.

A similar comparison between the out-of-phase tests performed at 121°C - 371°C (AT=250°C) and 121°C - 538°C (AT=417°C) is shown in Figure 10. As shown in this figure, the strain range remained relatively unchanged throughout low AT (250°C) out-of-phase tests as compared to the high AT (417°C) test. To explain the strain accumulation observed in high AT out-of-phase testing, it is postulated that progressive deformation occurred in matrix. In this system, progressive growth can occur if the load sharing capability of the matrix is reduced due to environmental damage (surface oxidation) or reversed plastic strain fatigue. At these high temperatures, the strain of the composite is limited by the presence of the stiff fibers. Therefore, the matrix can accumulate strain only to the extent limited by the fiber strain, and the load previously carried by the matrix must now be accommodated by the fibers. Finally, failure occurs when the strain limit of the fibers is reached.

In general, several major damage accumulation mechanisms were observed for the TMF tests conducted. Environmental damage (surface oxidation), reversed plastic strain fatigue of the matrix material and cyclic plastic ratcheting appear to be the dominant factors that influence failure in these metal matrix composites.
mechanisms, can be mapped on a diagram which matrix and interface form a complex picture of. This type of damage map can be used to establish material fatigue resistance. The damage accumulation mechanisms in fibers, which has coordinate axes taken as damage parameter, different in separable regions. The growth rates of the individual damage mechanisms might be found to be quit together. The growth of the composite and help to develop a life prediction model. For monolithic materials, such fracture mechanism maps have been developed, as shown in Reference (10), and have then been used to evaluate the fatigue performance of continuous fiber reinforced polymeric matrix materials (11). Recently, a similar map was developed for SiC fiber-reinforced titanium alloys (9).

In this study, the thermomechanical fatigue behavior of the SiC/Ti-6-4 composite system was classified using three parameters: 1) strain range increase rate (SRIR) (which is defined as percent of strain range change per cycle), 2) imposcred cyclic temperature range (ICTR), and 3) crossover temperature (which is defined as the most damaging combination of stress and temperature relative to matrix material yield stress). The resulting fatigue-life diagram is shown in Figure 11, where the imposed cyclic temperature range and crossover temperature are plotted against the logarithm of the number of cycles to failure. As seen in the figure, the failure mechanism for the out-of-phase cycle is dependent upon the cycle temperature range. Below a ΔT of 260°C - 288°C, failure of the composite is controlled by the fiber interface. Thermal cycling produces micro-damage associated with the matrix-fiber interface, namely microvoids or cracks, often leading to the debonding of the interface. The sequential application of an external load causes propagation of the crack in the matrix. For tests above cyclic temperature range of 288°C, all damage occurred in the matrix.

In Figure 12, a fatigue-life diagram was constructed using cycling temperature range versus the logarithm of SKIR. As seen in this figure, the classification of failure is in agreement with the above observation. The fatigue-life diagram in which the coordinates are the SRIR and the logarithm of the number of cycles to failure, is shown in Figure 13.

MICROMECHANICAL MODELLING

In order to provide more quantitatively accurate results to the investigation, a detailed numerical solution using the finite element technique is presented. The model to be examined here will be based upon the assumption of a regular square array of fibers. A section of one quarter of a fiber corresponding to the composite material was modelled using generalized plane strain elements. The stress-strain response of the composite system under thermomechanical loads was predicted using the ABAQUS nonlinear finite element analysis program (13).

Boundary conditions consistent with the assumption of a unit square-cell can be applied to the finite element model shown in Figure 14. Temperature dependent material properties for the Ti-6-4 matrix are listed in Table II. These properties were obtained by degrading sheet properties of Ti-6-4 due to the hot isostatic press process used to consolidate the composite at 899°C.

The nonlinear behavior of Ti-6-4 was modelled using plasticity theory based on a Mises yield surface which can expand isotropically and translate kinematically in stress space. The stress-strain response of fiber and matrix for lower temperature ranges is plotted in Figure 15. As seen in this figure, the matrix yields when the fiber expanded and contracted elastically during the cyclic properties when the yield surface is crossed following the first stress reversal.

Massive surface cracking was observed at cyclic high temperature ranges because of the gross oxidation effects. This resulted in degradation of the matrix strength during cycling. The material strength change was modelled using ORNL constitutive theory in ABAQUS (14). The ORNL plasticity theory provides the step change in the size of the yield surface from the virgin properties to the cyclic properties when the yield surface is crossed following the first stress reversal.

The stress-strain response of the composite under the high temperature cyclic range is shown in Figure 16. Again, the fiber remained elastic and the matrix yielded alternately in tension and compression as the temperature and load were cyclic under this loading. Certain conditions should occur, and when they do, produce a ratcheting growth of the composite. Since the fiber is always elastic at the test temperatures, strain range increase can happen only if the matrix losses its strength during cycling. This behavior is clearly demonstrated in Figure 16, as indicated by the points labeled B and D.
DISCUSSION

From the TMF life data and damage analysis at different applied temperature range, the fatigue damage maps for the SiC(SCS-6)/Ti-6-4 composite system, were constructed. Similar fatigue damage maps were constructed for the fiber reinforced polymer and titanium and titanium aluminide matrix composites under isothermal loading (11,9). The diagrams were plotted as applied maximum strain in composite vs. the fatigue life in Reference 11. The maximum applied stress in the fiber was used to construct the fatigue damage map for titanium and titanium aluminide based composite systems. In this study, to establish the physical understanding of damage modes and to quantify the damage accumulation rates of the mechanisms, the macrostructural and test parameters were used to construct the fatigue damage maps, as shown in Figures 11, 12 and 13. It was observed that the damage accumulation mechanisms was controlled by macrostructural parameters. Therefore, the applied maximum stress in the fiber was not used in the damage map construction. Based on the fatigue damage map and macrostructural observation, the TMF damage of the SiC(SCS-6)/Ti-6-4 composite system can be classified into two regions: 1) interface, matrix cracking and fiber breakage, and 2) matrix cracking dominated. Because of the design usable temperature range and stress level used for the TMF evaluation, fiber breakage dominated failure was not observed.

A fatigue damage map was used to evaluate the fatigue performance of several composite systems under isothermal condition (9). As far as the author is aware, no study involving the investigation of the effect of temperature cycling range on Ti-6-4 was based on composite system. Other research effects are being conducted to investigate damage mechanisms in this class of material under TMF loading(6). In Reference (9), the TMF behavior of MMC SiC(SCS-6)/Ti-15-3 was investigated at various maximum stress levels under both in-phase and out-of-phase conditions. The maximum applied stress was in the range of 827 to 1241 MPa and the temperature cyclic between 93°C and 635°C. On the stress range basis, in-phase fatigue lifes were found to be shorter than out-of-phase fatigue life. The authors concluded that the out-of-phase fatigue life controlled by matrix dominated failure, in-phase by fiber dominated failure.

In this study, the out-of-phase TMF cycle proved to be somewhat more damaging than the comparable in-phase and isothermal cycles (Figure 17). However, comparison with the in-phase data is not possible because of the limited number of tests performed. One specimen showed longer life than out-of-phase and isothermal test at 371°C maximum temperature and was terminated. A second specimen fractured prematurely when tested at a higher stress.

SUMMARY AND CONCLUSIONS

1. The damage mechanisms of the SiC/Ti-6-4 composite system under thermomechanical loading was identified by developing fatigue-life diagrams. Strain range change rate, cyclic temperature range and crossover temperature were proven as useful parameters to measure the fatigue resistance of this composite system for thermomechanical loading. These evaluations indicated that the failure mechanism for the out-of-phase cycle is dependent upon the cycle temperature range. Below a AT of about 260°C - 280°C, failure of the composite is controlled by the fiber interface. Above a temperature of 280°C, matrix dominated failure occurred.

2. A detailed numerical solution of the square-cell fiber geometry was performed to provide a more quantitative explanation of the failure mechanisms. The characteristics of the ratchet-growth mechanisms were developed using a finite element technique. There are several potential sources of progressive deformation as a result of temperature cycling. The occurrence of progressive distortion was analytically demonstrated and explained. The progressive distortion is attributed to the degradation of the matrix properties due to the observed multiple crack initiation sites on the surface.

3. It was observed that the effect of oxidation is a key factor leading to structural material damage under thermomechanical fatigue loading. This limits the application of Ti-6-4 based composite system above the maximum temperature of 250°C. Below this temperature, the reversed plastic strain of matrix material can be precluded using low module and high creep resistance material.
REFERENCES


12. ABAQUS, Hibbitt, Karlsson and Sorensen Inc., 100 Medway St., Providence, Rhode Island, 02906, U.S.A.

Figure 1  TMF Cycles in this Investigation.

Figure 2  Specimen Detail.

Figure 3  Fracture Surface of Tensile Specimen Tested at Room Temperature,
Figure 4  Fracture Surface of Isothermal Fatigue Specimen Tested at 371°C.

a) Fiber Pullout

b) Ductile Dimpling of Matrix

c) Crack Region
Figure 5  Fracture Surface of TMF Fatigue Specimen Tested at Low ΔT (250°C),

a) Fiber Pullout
b) Ductile Dimpling of Matrix
c) Crack Region
d) Surface Crack
Figure 6  Fracture Surface of TMF Fatigue Specimen Tested at High ΔT (528°C).

a) Multiple Surface Crack Origination
b) Multiple Origin on Different Plane
c) Ductile Dimpling of Matrix
Figure 7  Strain History of Out-of-phase Test with Temperature Extremes of 121°C and 454°C. The strain range and mean strain values increased as cycles accumulated.

Figure 8  Strain History of In-phase Test with Temperature Extremes of 121°C and 371°C. The strain range remained relatively unchanged while the mean strain increased in the early stages of the test.

Figure 9  Strain History Comparison Between In-Phase and Out-of-Phase at 121°C - 371°C TMF Tests.

Figure 10  Strain History Comparison Between 121°C - 371°C and 121°C - 538°C Out-of-Phase TMF tests. The strain range remained relatively unchanged throughout low ΔT test as compared to high ΔT test.

Figure 11  Fatigue Damage Map Based on Variables, Imposed Cycling Temperature Range and Crossover Temperature vs. Number of Cyclic to Failure.
FAILURE MODE
MASSIVE SURFACE CRACKING

FAILURE MODE
INTERFACIAL DAMAGE
MATRIX DUCTILE FAILURE
FIBER PULLOUT

Figure 12  Fatigue Damage Map Based on Variable, Imposed Cycling Temperature Range vs. Strain Range Increase Rate.

Figure 13  Fatigue Damage Map Based on Variable, Strain Range Increase Rate vs. Number of Cycles to Failure.

Figure 14  Finite Element Model of Square-Cell Fiber-Matrix.

Figure 15  Stress-Strain Response of Fiber and Matrix to Lower Temperature Range Out-of-Phase TMF Cycling.

Figure 16  Stress-Strain Response of Fiber and Matrix to High Temperature Range Out-of-Phase TMF Cycling.
Comparison of TMF Live with Isothermal Fatigue Live. The out-of-phase TMF cycle proved to be somewhat more damaging than the comparable in-phase and isothermal cycles.

**Figure 17**

**TABLE 1**

**TEST MATRIX**

<table>
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<tr>
<th>TMF PHASE</th>
<th>TEMPERATURE (°C)</th>
<th>STRESS (MPa)</th>
<th>DELTA TEMP</th>
<th>T MAX</th>
<th>T MIN</th>
<th>MAX</th>
<th>MIN</th>
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<tr>
<td>IN-PHASE</td>
<td>250°C 370°C 120°C</td>
<td>144.0</td>
<td>15.7</td>
<td>144.0</td>
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<tr>
<td>IN-PHASE</td>
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<td>198.0</td>
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<td>198.0</td>
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<td>171.0</td>
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<td>OUT-OF-PHASE</td>
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**TABLE II**

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<tr>
<th>TEMPERATURE (°C)</th>
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<th>ULTIMATE STRENGTH (MPa)</th>
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