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Ambient Temperature Strength Degradation of a Ceramic Matrix Composite Due to Solid Particle Erosion

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Honors Research Project – Bachelor of Science in Mechanical Engineering

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Abstract

Solid particle erosion (SPE) is an issue for aircraft exposed to hard, fine particles, such as sand or dust, that can be ingested into the engine at a high velocity without reaching their melting point, resulting in material removal. SPE can be attributed to the intake of particulate debris from the runway or from airborne particulates. The cumulative mass loss from SPE damage affects the structural integrity and performance of the engine; however, its effects on ceramic matrix composites (CMCs), that are being employed in the aerospace industry, are not well understood in literature. The limited SPE research studies that have been published are primarily based on woven silicon carbide (SiC)-based CMC architectures. Therefore, the aim of this study was to characterize the effects of ambient temperature SPE damage on the mechanical properties of a SiC-based cross-ply laminate CMC system. The effects of erodent particle velocity and cumulative erodent particle mass were quantified with respective to the damage morphology and retained tensile strength.
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Introduction

The material science field is generally classified into six distinct groups: metals, ceramics, glasses, polymers, elastomers, and hybrids. Composites are a class within the hybrid group and refer to a material produced from two or more constituent materials. These constituent materials have notably dissimilar chemical or physical properties that contribute to the final material properties. Ceramic matrix composites (CMCs) consist of a ceramic matrix reinforced by continuous ceramic fibers. The matrix holds the material together and serves to transfer load to the reinforcement fibers. Additionally, a CMC may include a weak fiber-matrix interphase, usually in the form of a coating, to improve toughness and enable graceful failure [1]. The development of CMCs has occurred over several decades, and their emergence can be traced back to the latter half of the 20th century [2]. Despite their history, applications remained scarce, especially in the realm of commercial markets, until the past decade when considerable progress has been made in the development of CMCs to replace conventional nickel-based superalloys that have long dominated the aerospace and gas turbine industries [3].

CMCs offer a variety of performance advantages over their metallic alloy counterparts, primarily a material density that is one-third that of the nickel-based superalloys currently in use. This translates to components 50 percent lighter in weight, accompanied by improved thermomechanical properties that reduce the need for cooling, enabling a fuel burn reduction of up to two percent [4]. This advanced capability allows gas turbine engines to operate more efficiently at higher temperatures, burning fuel more completely and emitting fewer pollutants [5]. For these reasons, future military aircraft engines will continue to adopt CMCs into their architectures to increase thrust, efficiency, and flight range. The successful integration of CMCs into military applications has also paved the way for their adoption in the commercial sector. Notably, engines like the GE9X, which will enter service on the Boeing 777X in 2025, exemplify this trend with the incorporation of five major CMC hot-section components [6].

Fig. 1 Leading-edge erosion damage observed on a high-pressure turbine blade from a NASA DC-8 engine [9].

Fig. 2 (A) A typical gas turbine engine intake vortex [7]; (B) Ash plume generated during the 2010 volcanic eruption of Eyjafjallajökull in Iceland [8].
The integration of CMCs into turbine engines exposes the material to harsh, erosive environments, especially in the military sector. Solid particle erosion (SPE) is a common phenomenon that occurs in aircraft gas turbine engines when particles that have not exceeded their melting point are ingested at a high velocity, resulting in material removal. This occurrence is particularly apparent along aerodynamic channels as illustrated in Fig. 1, showcasing erosion damage to the leading edge of a high-pressure turbine blade from a NASA DC-8 engine [9]. SPE can occur during aircraft taxiing or takeoff, arising from the ingestion of particles from the runway. Fig. 2(A) provides a visual representation, illustrating the gas turbine engine intake vortex that is formed during engine run-up [7]. Additionally, SPE can occur at an altitude as a result of airborne particles. Fig. 1(B) displays the expansive reach and impact of an airborne hazard arising from the 2010 volcanic eruption of Eyjafjallajökull in Iceland. This event caused significant disruptions in European air travel for nearly a week due to concerns of particle ingestion [8]. The cumulative mass loss from SPE damage to CMC components in gas turbine engines impacts both the structural integrity and performance of the engine. However, the influence of erosive environments on the mechanical properties and failure mechanisms of CMCs is not well understood in literature.

A pioneering study investigated the erosion behavior of gas turbine grade CMCs using a 230 mesh (63 μm) erodent particle size at a velocity of Mach 1, approximately 343 m/s, and compared the results to those of monolithic ceramics, thermal barrier coatings (TBCs), environmental barrier coatings (EBCs), and nickel-based superalloys [10]. The erosion rate of the CMCs was quantified with respect to nominal density, matrix hardness, and elastic modulus. The results highlight the susceptibility of CMCs to material removal and the complexity of the corresponding damage mechanisms. The study concluded that CMCs that possess a higher elastic modulus, density, and hardness exhibit greater erosion resistance. Since then, several studies have investigated the erosion behavior of melt-infiltrated SiC-based woven CMCs in response to a wide range of erodent particle sizes, velocities, impingement angles, and concentrations. The erosion of CMCs has been found to be a complex and iterative process governed by macroscopic material properties [10]. Experimental investigations have shown that the erosion rate increases with erodent particle size, velocity, impingement angle (maximum at 90°), and operating temperature [3,10-13]. The effect of temperature on erosion rate is attributed to a decrease in strength and hardness of the material. Conversely, erosion rate has been shown to decrease with an increase in erodent particle concentration attributed to a particle shielding effect [3]. The primary material removal mechanism in SiC-based CMCs is attributed to subsurface cracks under normal impact angles and a ductile ploughing mechanism, as described in literature, at shallow impingement angles [3].

The aim of this study was to characterize the effects of ambient temperature SPE on the mechanical properties of a melt-infiltrated SiC-based cross-ply laminate CMC that is representative of CMC systems in use today. The influence

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**Fig. 3 Melt-infiltrated cross-ply laminate composite microstructure at (A) low and (B) high magnification.**
different erosive exposures were investigated by varying erodent particle velocity and cumulative erodent particle mass. The resulting SPE damage was quantified with respect to the material mass loss and damage morphology. Additionally, the retained tensile strength of the material was measured, and the net-section tensile strength of the remaining material was determined with respect to the SPE damage. Non-destructive evaluation (NDE) techniques, including digital image correlation (DIC), electrical resistance (ER), and acoustic emission (AE), were coupled with optical microscopy analysis to investigate the damage accumulation and failure mechanisms as well as to explore material monitoring methods. The scope of this paper will be limited to discussing the results of the optical microscopy analysis.

**Experimental Methodology**

The CMC target material is a vintage GE Aviation prepreg laminate. The composite consists of individual unidirectional laminas in the form of a tape of continuous Hi-Nicalon™ Type S (HNS) fibers. In the prepreg process, the SiC fibers are coated with a boron nitride (BN)-based fiber-matrix interphase and a protective silicon nitride (Si₃N₄) overcoating using a proprietary chemical vapor deposition (CVD) process and then formed into unidirectional prepreg sheets via wet drum winding [1]. The prepreg sheets are then laid-up and laminated to form the composite preform. At this stage, the matrix of the composite consists of powders (commercially available SiC and C) in a mixed polymer binder. A binder burnout/pyrolysis step converts part of the polymer to carbon. Final densification takes place when molten silicon infiltrates the porous preform, during which the silicon reacts with the free carbon present in the preform to form a continuous SiC phase in the matrix [1].

The composite structure is comprised of eight plies that are oriented in a symmetrical cross-ply lay-up ([0/90]₄). The nominal volume fraction of the composite is approximately 20-25% [1]. Fig. 3 shows the typical microstructure of the composite at (A) low and (B) high magnification. The low-magnification image distinguishes the 0- and 90-degree fiber layups, and the high-magnification image highlights the fiber distribution within the matrix. The BN fiber-matrix interphase is differentiated by the bold outline of the fibers, whereas the light blue sections are the presence of free Si within the matrix. Despite the antiquated nature of the material, it is considered representative of a baseline melt-infiltrated laminate CMC system in use today.

The erodent material used in this study is 120 mesh (125 μm) commercially available white aluminum oxide particles obtained from Kramer Industries, Inc. As explained by Panakarajupally et al. [3], the alumina particles resemble a rhombohedral shape with sharp edges that are ideal for erosion testing. The erodent particles were sprayed at an impingement angle of 90° (normal) to the surface at two different velocity conditions, 200 and 350 m/s (Mach ≈ 0.6 and Mach ≈ 1.0, respectively) and two different cumulative erodent particle mass quantities, 1 g and 2 g.

The ambient temperature erosion procedure aligned with the setup in previous erosion studies [3,11-13]. The testing was conducted using The University of Akron pneumatic impact gun. The experimental setup, illustrated in Fig. 4, uses pressurized helium (He) gas to propel erodent particles through a 300 mm steel barrel before impacting the target material at a normal impingement angle. The standoff distance between the target material and the barrel was set at 15 mm with the barrel aligned to impact the center of the specimen. The helium reservoir was pressurized to 100 and 200 psi to achieve the target erodent particle velocities of 200 and 350 m/s, respectively. The erodent particle velocities were verified using a double disc method that has been commonly applied to erosion studies [18].
Fig. 5. Cumulative mass loss of each combination of erosive conditions.

Fig. 6. Steady-state erosion rate comparison between erosive conditions.

Fig. 7. Power law relationship between erodent particle velocity and erosion rate.

Fig. 8. Crater depth progression with increasing cumulative erodent mass.

Fig. 9. Axial cross-sectional area growth of the crater with respect to cumulative erodent mass.

Fig. 10. Transverse cross-sectional area growth of the crater with respect to cumulative erodent mass.
The target cumulative erodent particle mass was divided into five iterations. The first set, totaling 1 g, impacted 0.2 g of erodent particles per impact iteration. The second set, totaling 2 g, impacted 0.4 g of erodent material per impact iteration. Between iterations the mass loss of the specimen was measured, and the damage morphology was analyzed using a Keyence VHX 7000 optical microscope. SPE damage was quantified with respect to the crater depth, axial cross-sectional area, and transverse cross-sectional area. Each variation of erodent particle velocity and cumulative mass were tested twice, except for the minimum condition of 1 g of cumulative erodent at 200 m/s which was only tested once since it produced only negligible erosion damage compared to the average surface roughness of the specimen.

A dogbone specimen, characterized by a reduced cross-sectional area in the gage region, was used to collect baseline material properties whereas a straight-edge specimen was used to evaluate the retained strength at each
combination of erodent particle velocity and cumulative erodent particle mass. The average dogbone specimen dimensions were 152 mm × 10.2 mm × 1.8 mm. The average straight-sided specimen dimensions were 152 mm × 12.7 mm × 1.8 mm. All specimens were machined from three distinct panels of the target material. The baseline material properties and retained strength of the eroded specimens were both evaluated by an ambient temperature monotonic tensile test. This involved applying uniaxial loading to the specimen at a displacement rate of 0.15 mm/min, continuously deforming it until failure. An extensometer, with a standard 25.4 mm (one inch) gage region, collected strain data for the baseline dogbone specimens. Conversely, digital image correlation was employed to collect strain data on the pristine backside of the straight-sided eroded specimens. AE and ER NDE monitoring techniques were applied to both the baseline and eroded tensile tests; however, the results will not be presented or discussed in this paper.

**Results & Discussion**

**Erosion Behavior**

The influence of the erodent particle velocity and cumulative erodent particle mass on the erosion behavior of a melt-infiltrated SiC-based cross-ply laminate CMC system was explored at a normal impingement angle under ambient temperature. Fig. 5 highlights the cumulative linear mass loss of the specimen material with respect to cumulative erodent particle mass. The legend indicates distinct material panels by their respective identification numbers (e.g., 0025, 0028-09, and 0028-10). The linear progression and general agreement between specimens tested under the same conditions verifies the consistency of the test setup.

The effect of the erosive conditions is quantified in terms of the erosion rate on a mass basis as defined in equation (1) where \( E \) is the erosion rate, \( m_i \) and \( m_f \) are the initial and final mass quantities of the target material, and \( m_e \) is the mass of the erodent material.

\[
E = \frac{(m_i-m_f)}{m_e} \tag{1}
\]

An erosion rate comparison is shown in Fig. 6 for the various erosive conditions. The erodent particle velocity and cumulative erodent particle mass both had a similar influence on the erosion rate of the CMC. For a constant cumulative erodent particle mass, increasing the erodent particle velocity from 200 m/s to 350 m/s at 1 g of total erodent particle mass resulted in an approximately a fourfold increase in the erosion rate. Conversely, when the same increase in velocity occurred at 2 g of total erodent particle mass, it led to only an approximate threefold increase in erosion rate. This indicates that the influence of velocity on the erosion rate was higher at the lower cumulative erodent particle mass quantity. In a similar manner, holding erodent particle velocity constant, an increase in cumulative erodent particle mass at the lower velocity resulted in nearly a one-fold increase in the erosion rate, compared to the effect at the higher velocity.

In literature [15-17], it is reported that the steady-state erosion rate and erodent particle velocity follow a power law relationship for monolithic ceramics of the form shown in equation (2) where \( V \) is velocity, \( a \) is a constant coefficient, and \( n \) is the velocity exponent that typically ranges from 2 to 4 [11].

\[
E = aV^n \tag{2}
\]

An analysis of quasi-static indentation fractures resulting from sharp particle impacts suggests a velocity exponent of 2.33 [18] or 2.44[19] when lateral cracking dominates the material removal process. In contrast, a dynamic analysis predicts a velocity exponent of 3.17 [20]. Fig. 7 illustrates the erosion rate as a function of velocity, along with the respective power law velocity exponent values. Despite the presence of fibers, the velocity exponents obtained are consistent with those of monolithic ceramics where subsurface lateral cracks are the predominant driver for material removal.

Figs. 8-10 characterize the damage morphology through analysis of the crater depth, axial cross-sectional area, and transverse cross-sectional area in relation to cumulative erodent mass. A linear relationship is observed between the progressive crater growth and cumulative erodent mass with general agreement among the results from different panels. The similarity between the axial and transverse cross-sectional crater areas indicates the crater topography closely resembles a hemisphere. Fig. 11 illustrates the optical three-dimensional (3D) scan of the crater morphology. This approach not only enhanced the accuracy of estimating the net section strength of the target material, but also provides valuable insight into the fluctuation of erosion resistance between composite layers. Compared to a melt-infiltrated SiC-based woven CMC tested under identical conditions in the same facility, the melt-infiltrated cross-ply
lamine CMC exhibited greater resistance to erosion damage, resulting in a crater depth only two-thirds that of the woven counterpart [4].

**Material Strength**

To assess the material properties of the target material, particularly tensile strength, three dogbone specimens underwent monotonic tensile testing. Additionally, cyclic tensile loading was applied to one dogbone specimen to gain insights into material damage accumulation, tracked through AE data (details not discussed in this paper).

The retained tensile strength of the eroded specimens was evaluated by a monotonic tensile test. Fig. 12 depicts the engineering tensile strength degradation of the material due to SPE damage. Given the proprietary nature of the material performance, the y-axis has been hidden to conceal the exact engineering tensile strength values. Notably, even under the most severe erosive conditions, with the crater depth exceeding half the thickness of the specimen, there was only a minor amount of degradation in engineering tensile strength. Net section tensile strength, which accounts for the cross-sectional area reduction at the crater, was calculated to further assess and compare the strength degradation of the eroded specimens. The values have been normalized to examine the impact of the crater, functioning as a stress concentrator, on the degradation of strength. This involved dividing both the baseline and retained tensile strength values by the average baseline tensile strength obtained from the pristine dogbone specimens. The results in Fig. 13 show that despite the presence of a crater, there is no net stress concentration effect due to the crater.

**Conclusion**

The erosion behavior and retained tensile strength of a melt-infiltrated SiC-based cross-ply laminate CMC system were investigated with a focus on erodent particle velocity and cumulative erodent particle mass. The velocity exponents obtained support the conclusions of similar CMC erosion studies that identified subsurface lateral cracks as the primary material removal mechanism similar to the behavior of monolithic ceramics despite the presence of fibers. The SPE damage morphology was characterized by crater depth, axial cross-sectional area, and transverse cross-sectional area and compared to a woven CMC counterpart. The laminate exhibited an increase in erosion resistance characterized by a 33% reduction in crater depth under the same erosive conditions. The transverse cross-sectional area of the crater was used to accurately calculate the retained net section strength of the material under monotonic tensile loading. The results indicate that although the material experiences degradation in engineering strength, the net section strength remains unaffected. The resilience of the CMC under erosive conditions is an encouraging sign for further expansion of CMCs into gas turbine engines. It's important to note that the erodent amounts used in this study were relatively small compared to real-world situations such as exposure to large volcanic ash plumes or continuous operation in sandy environments. Additionally, CMCs typically endure elevated temperatures in many applications, which can compromise their erosion resistance. Future work will aim to connect ER, AE, and DIC data to physical damage mechanisms related to SPE damage. Subsequent investigations will explore the impact of elevated temperatures on the retained strength in a simulated combustion environment to better understand the erosion response of laminate CMC systems under turbine engine operating conditions.

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