Experimental Study of Thermo-Mechanical Behavior of SiC Composite Tubing under High Temperature Gradient Using Solid Surrogate

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1 Abstract

Nuclear grade silicon carbide fiber (SiC_f) reinforced silicon carbide matrix (SiC_m) composite is a promising candidate material for accident tolerance fuel (ATF) cladding. A major challenge is ensuring the mechanical robustness of the ceramic cladding under accident conditions. In this work the high temperature mechanical response of a SiC_f-SiC_m composite tubing is studied using a novel thermo-mechanical test method. A solid surrogate tube is placed within and bonded to the SiC_f-SiC_m sample tube using a ceramic adhesive. The bonded tube pair is heated from the center using a ceramic glower. During testing, the outer surface temperature of the SiC sample tube rises up to 1274 K, and a steep temperature gradient develops through the thickness of the tube pair. Due to CTE mismatch and the temperature gradient, the solid surrogate tube induces high tensile stress in the SiC sample. During testing, 3D digital image correlation (DIC) method is used to map the strains on the outer surface of the SiC-composite, and acoustic emissions (AE) are monitored to detect the onset and progress of material damage. The thermo-mechanical

Abbreviations:

E = 10; CTE = coefficient of thermal expansion; RT = room temperature

behavior of SiC-composite sample is compared with that of monolithic SiC samples. Finite element models are developed to estimate stress-strain distribution within the tube assembly. Model predicted surface strain matches the measured surface strain using the DIC method. AE activities indicated a progressive damage process for SiC_f-SiC_m composite samples. For the composites tested in this study, the threshold mechanical hoop strain for matrix micro-cracking to initiate in SiC_f-SiC_m sample is found to be ~300 microstrain.

2 Introduction

Ceramic composites consisting of nuclear grade, highly stoichiometric, silicon carbide fiber and matrix (SiC_f/SiC_m) are being developed as the next generation accident tolerant fuel cladding for light water reactors (LWR) [1, 2]. The nuclear fuel cladding tube is designed to contain the fuel pellets and to prevent the release of radioactive fission products into the coolant and eventually to the outside environment [3]. The toughness of ceramic composites under thermo-mechanical loading needs to be evaluated not only under normal operation conditions but also in the accident conditions such as the loss of coolant accident (LOCA) and the reactivity initiated accident (RIA). Therefore, it is of great interests to develop practical methods and capabilities of replicating extreme thermo-mechanical conditions potentially encountered by the cladding during severe accidents to help understand the failure modes and failure resistance of the ceramic cladding.

In a typical light water reactor, the fuel cladding tube experiences a uniform differential pressure loading from coolant outside and gas (Helium and fission gas) inside. The net pressure is compressive during normal operation. Mechanical contact between the cladding and fuel pellets can occur due to the creep of the cladding and the radiation-induced swelling of fuel

pellets during service, this will result in detrimental localized high mechanical contact stress which could lead to mechanical failure [4, 5, 6]. During LOCA and RIA, the fuel cladding develop excessive stress and deformation, due to loss of coolant pressure, rising internal gas pressure, severe pellet cladding mechanical interaction (PCMI), high temperature, and high thermal gradients through the thickness of the cladding [6,7]. Behavior of Zircaloy cladding under accident conditions has been extensively studied [8 to 17]. However, relatively little is known about the behavior of $\text{SiC}_r\text{-SiC}_m$ cladding under severe accident conditions. It is of great importance to develop better understanding of thermal-mechanical response of ceramic claddings during accident scenarios. This is the motivation of the work reported in this paper.

A number of mechanical testing and evaluation methods have been applied to nuclear fuel cladding and cladding materials. They can be classified into two general categories: tests for generating fundamental material properties data; and tests for mimicking certain aspect of service condition with the purpose of material screening or proof-testing the component. The first class of tests are typically conducted using relatively small sample under isothermal conditions. These include O-ring tension test, close-end pressure burst test, expansion-due-to-compression test, and C-ring compression ring. The O-ring tension test and internal pressure burst test can be conducted at high-temperature.

Chang-Sung Seok et al. [8] applied ring tension test and burst test to a cladding sample made of Zirconium at high temperatures between 639 to 844 K. The test features a heating chamber where the Zirconium ring specimen is held between two half cylinders that apply the load with a central piece inserted in the ring to reduce the bending deformation. The ring tensile test requires numerical analysis to calculate the stresses and assumption of the friction between specimen and mandrels [9]. The hard contact with half-mandrel or half-cylinder inevitably introduces some

stress concentration [4]. All of the studies cited above were conducted in nearly iso-thermal conditions.

Internal pressure tests of burst tests of tubular specimens can be achieved by liquid (water, oil), gas (air, nitrogen, argon, helium), or solid (silicon rubber, aluminum, copper) [9,10] pressurizing media. The choice is largely dictated by the desired temperature range required for the test. Reported burst tests of nuclear fuel cladding typically use oil or gas as the pressurizing media. Close-end burst tests of nuclear fuel cladding at high temperature are preferentially used compared to the open-end burst test. This is due to the fact that closed end test creates a biaxial stress state which is similar to the stresses developed in a LOCA. A $\sigma_{hoop}/\sigma_{axial} \sim 2$ occurs when the ends of the specimen are closed, and a uniaxial state of stress, σ_{hoop} , is developed when the ends are open or not constrained [9, 7]. High temperature and high pressure burst behavior of Zircaloy has been extensively reported. Yamanaka et al [11] studied the fracture behavior of a hydride Zircaloy cladding tube using a close-end burst test. The sample was heated up to at a constant temperature of 621 K and then hydraulically pressurized at the rate of 2 MPa/s until failure. Reported burst pressures are between 77 and 126 MPa. Similar work was conducted by Nagase et al. [12] where a hydrides Zircaloy cladding specimen was heated to 621 K and then hydraulically pressurized at a maximum rate of 0.2 GPa/s until failure.

Safety and cost are major concerns when using liquid or gas in high temperature burst test. The rapid release of stored energy from compressed gas (several thousands to 10s of thousands Psi) can results in damage to the furnaces, instrument, or even the operator. The spray of very high temperature oil, liquid metal or glass during sample burst may cause fire, damage to equipment and injury to operator. Expensive high pressure pumps are typically needed and sealing for high temperature burst test is challenging. Very long sample tubes are often needed

to allow at least one end extending to the low temperature zone for sealing. Alternatively, solid phase material can be used in expansion-due-to-compression (EDC) experiment to generate internal pressure. An axially compressive load is used to displace a solid pellet inserted in a cylindrical specimen (cladding) causing the deformation of the pellet, thus inducing pressure on the inner surface of the cladding. EDC testing has been performed with a PTFE (Polytetrafluoroethylene) pellets for temperatures of less than 604 K; and soft metals like aluminum or copper pellets [9] for higher temperature. The practical usable temperature for EDC methods is not high enough to simulate severe accident condition of ATF cladding. In addition to temperature limitations, numerical analysis is sometime necessary to calculate stresses for the EDC test, particularly when significant friction force exists between the contact faces or for non-uniform materials. These types of burst tests require larger specimens compared to specimen for C-ring and O-ring tests.

In the second class of the mechanical study, the cladding samples are exposed to some adverse service condition and their mechanical responses are observed. Hardy [13] conducted a transient temperature test of Zircaloy-4 cladding to study the effects of internal pressure and heating rate in the expansion of the cladding through rupture simulating a LOCA. The tests were performed at transient temperature with constant pressure, and isothermal condition with rising pressure. The range of temperatures was from RT to 1148 K, and pressure range was 0.34 to 13.8 MPa; the pressurization media was helium. Similar work conducted in steam environment were reported by Sagat et al. [14] and Hindle et al [15] to study the deformation of the cladding exposed to an oxidizing environment. An open-end burst test of a Zircaloy cladding at a temperature of ≤ 624 K and maximum pressure of 100 MPa was reported by Nakatsuka et al.

[16]. With few exceptions, Zircaloy develops large plastic deformation (bulging) during high temperature burst test and the failure pressure reduces rapidly with temperature.

Studying the mechanical response of ceramic based cladding material is more challenging, primarily due to the high elastic modulus of ceramic materials and difficulties in quantifying microscopic damage initiation and accumulation process in ceramic composites. Fundamental material property data of ceramics are typically obtained by 3 or 4-point bending test of small coupons and burst test of small tubular samples

Srinivasan [17] studied the stress rupture response of sintered α -silicon carbide at elevated temperature using a 4-point bend test. Strength of the sample was found to gradually degrade over time at temperatures of 1623 K and higher in air. At 1623 K the stress rupture decreases in about 20%, from 400 to 500 MPa, in 10 hours. Shelleman et al [18] reported high temperature internal pressure test of silicon carbide tubes tested in an explosion-proof chamber with Argon or air as pressurizing media. A temperature of 1673 K and a burst pressure of 68.9 MPa was reported.

Chuck et al. [19] reported an interesting high-temperature internal pressure experiment that used glass as the pressurizing media. Glass deforms viscoelastically above its softening temperature. This method was used to test a ceramic matrix composite tube (braided Nicalon/PIP Si-C-N) at the temperature of 1274 K in vacuum, using induction heating to heat the specimen and glass. As temperature rises, the viscosity of the glass decreases rapidly and can behave like a hydraulic fluid. However, this method requires a high temperature end seal. This practically limits the upper usable temperature of this method.

A miniaturized fracture stress test of thin-walled tubular SiC specimens at room temperature was reported by Byun et al [20]. The internal pressure was applied using the rubber plug technique and the fracture stress calculated was 238.8 MPa. The tubes were made by CVD of SiC on pyrolytic carbon, having I.D. of 1mm. More recently, the test and simulation of a 3-layer silicon carbide cladding under simulated PWR conditions has been reported by Stempien [21] and Lee [22]. The cladding is composed of three layers (Triplex): an inner monolithic SiC layer, a central SiC/SiC composite layer, and an outer environmental barrier coating (EBC). The monolith is manufactured via CVD process and the composite is made of Hi-Nicalon type S SiC fibers wrap around the central monolith with a matrix deposited by CVI process. The EBC is a thin (50 – 150 μm) SiC coating applied to the outside of the composite. Stempien [21] studied the mechanical behavior of irradiated Triplex SiC specimens at room temperature using the rubber plug technique.

Very few studies in the literature report the mechanical response of ceramic cladding under simulated accident conditions. Lee et al [22, 23] studied the performance of sintered α -SiC tubes and CVD β -SiC bars under thermal shock (quenching test). During the thermal shock the specimens were heated in a furnace until they reached a constant temperature (up to 1533 K), then the hot specimen was quickly dropped into a water pool at room temperature. The strength of the specimens after thermal shock was obtained by the rubber plug technique for the sintered α -SiC and 4-point bending method for the β -SiC. Another thermal shock experiment of a fiber reinforced ceramic matrix composite has been reported by Eckel et al [24]. The specimen was composed of plain woven Nicalon fibers in a 2-D laminated form and CVI matrix. The thermal shock was induced by exposing the specimen to hot combustion gas (in a turbine engine) with

temperatures between 1573 and 2773 K for 1 second and then was rapidly cooled down. The composite showed a decrease in tensile strength after 51 one-second thermal shocks at 1973 K.

In this paper, we apply a novel high temperature test method to evaluate the thermo-mechanical robustness of ceramic composite and monolith tubes under high temperature, high temperature gradient, and high mechanical stress simultaneously. The experimental scheme attempts to simulate some aspects of severe accident conditions (such as RIA) that include high temperature, a temperature gradient through the thickness, and hard PCMI. State-of-the-art strain measurement and acoustic emission monitoring techniques were applied during the high temperature test. Digital image correlation (DIC) method with a special light filtering technique was used to measure the surface strain distribution on hot sample; the surface strain distribution allows the direct calculation of surface stress in the elastic region, and to calibrate stress-strain models. AE monitoring was achieved using a long ceramic surrogate tube which extend out of the hot test zone, and transmit AE signals to a piezoelectric sensor outside the hot zone. The study allowed us to observe, for the first time, thermo-mechanical response of a $\text{SiC}_f\text{-SiC}_m$ composite tube under simulated accident conditions. To serve as control to provide contrast of the material behavior, monolithic SiC tubes were also tested using the same procedure. Finite element models were developed to help reveal the internal stress-strain distribution in the specimen during testing.

3 Materials and Methods

The objective is to observe thermo-mechanical behavior of SiC ceramic composite tube (the specimen) under internal pressure, at high temperatures and temperature gradients. Two types of sample materials were tested. One is a control sample made of monolithic SiC (Ortech Advanced

Ceramics), and the other is SiC_f-SiC_m composite tubing. The SiC_f-SiC_m composite tubing was fabricated at General Atomics from nuclear grade HI Nicalon Type S fiber preforms using a chemical vapor infiltration (CVI) processing route [9]. A slightly axially biased fiber architecture was used for the specimens tested. A thin pyrolytic carbon interphase was applied to the fibers via CVD. The SiC matrix was deposited via decomposition of methyltrichlorosilane. Composites were infiltrated to 88% theoretical density and fiber fraction in the range of 30-35%.

The experimental approach is shown in **Fig 1a**. A short segment of SiC sample tube is bonded to a solid surrogate tube made of alumina with a thin layer of ceramic adhesive in between. The tube pair is heated from the centerline of the assembly using a ceramic glower. Intense heating results in high temperature and a steep temperature across the assembly. The surrogate tube has a higher CTE than SiC sample and is heated to higher temperature, hence it will exert normal (pressure) and shear stress (**Fig 1c**.) to the SiC sample tube through the thin ceramic adhesive layer, resulting in a biaxial state of stress in the center portion of the SiC specimen. This approach allows us to induce temperatures over 1200 K, temperature gradient up to 250 K across the test assembly, and mechanical internal burst pressures, similar to conditions that might be expected during accident condition.

Figure1b. shows the test chamber designed for the high temperature experiment. The quartz windows allow the acquisition of images using CCD cameras for posterior strain analysis by DIC. The chamber has openings to accommodate instrumentation such as thermocouples and electrical leads. The heater is inserted from the top of the chamber into the specimen, and holes on the floor of the chamber are used to control the heat loss rate, by regulating an air flow around the specimen.

Figure 1.

An alumina tube (Al_2O_3 , AL-960, 96 % purity) is used as the surrogate and an yttria stabilized zirconia (YSZ, partially stabilized) rod operated in the Nernst glower [25] mode is used as a heater. Both are acquired from Ortech Advanced Ceramics. The ends of YSZ rod (O.D. 3.48 mm and 28 mm long) are wrapped with Pt-30Rh wires to form two electrodes which are connected to a variable voltage AC power source. At room temperature, the electrical impedance of YSZ is very high (on the order of several mega Ohm), but it reduces rapidly (to ~ 220 Ohm at ~ 1273 K) as temperature rises. Regular 300 W incandescent bulbs are put in series with the YSZ glower and serve as ballast. To start the glower, it is first preheated (to about 873 K) using either a torch or an electrical heater, and then a gradually increasing AC voltage was applied until the rod glows in a self-sustained manner. This voltage is the minimum voltage needed to “fire” the rod. The glowing rod has a minimum temperature of approximately ~ 1274 K. The softening and melting of the connecting wire (Pt-30Rh) limits the upper operating temperature of the glower. The glower assembly can operate stably in excess of 1974 K in air.

It is desirable to capture the surface strain distribution of the specimen during the thermo-mechanical “loading” process. Surface strain data can be used to calculate surface stress in the elastic range if stress-strain data at the corresponding temperature are available from more traditional tensile testing. In addition, the strain data can be used to validate computation models.

Digital Image Correlation (DIC) was used to capture sample surface strain during thermomechanical testing. It is a non-contact full-field strain measurement technique with broad applications [26]. Three-dimensional DIC method enabled by two cameras has been widely used

to capture strain distribution on curved surface. However, high temperature environments present several challenges to the application of DIC. First, a surface speckle pattern that can survive the high temperature exposure is needed. Second, hot material emits visible light which can change the contrast of the speckle image and confuse the DIC algorithm. Third, the hot air surrounding the sample may potentially distort the image. In this work, several techniques were applied to overcome/avoid these difficulties.

Figure 2 shows the stress-strain behavior of nuclear grade $\text{SiC}_r\text{-SiC}_m$ composite tube obtained at room temperature in the authors' lab. The result was obtained by internal pressure experiment on a tubular sample using both strain gages and DIC to capture the deformation of the sample. The curve shows the calculated hoop stress v.s. measured hoop strain on the surface (OD) of the sample. The hoop stress is calculated based on the applied internal pressure and thick wall cylinder theory. The standard deviation of the DIC strains, an indicator of the DIC measurement error, is also given in **Fig 2**. The error is well within the typical DIC measurement error of $200\ \mu\epsilon$ [28]. A distinct feature is a “knee” where stress-strain curve bends. It is a result of extensive matrix micro-cracking when the stress reaches the proportional limit stress (PLS). At the OD, the stress at the “knee” for the $\text{SiC}_r\text{-SiC}_m$ composite tube is approximately 90 MPa; the hoop stress at the ID at this pressure, where micro-cracking is expected to initiate, is 135 MPa.

Figure 2.

Through trial-and-error, we have found a commercially available Alumina spray paint (A-aerosol from ZYP-Coatings Company) which is capable of producing temperature resistance speckle pattern on the SiC sample surfaces. Chemical thermodynamic calculation indicates that

the SiC-Al₂O₃ interface is stable at temperature below 1800 K [27]. To overcome the second problem of high temperature DIC measurement, a light filtering scheme (Novak et al. [28]) is used. This approach illuminate and image the sample through a narrow spectral window around 480 nm, thus avoided thermal emitted photons mostly at longer wavelength. This arrangement allows the capture of images with stable surface contrast during the heating process [29].

The distortion of the images due to changes in the refraction index of hot air, known as heat haze, is another concern when images are acquired at high temperature. The worst case occurs when viewing through turbulent hot air plume or when the viewing direction is perpendicular to the temperature gradient, as in hot-road mirage. A sheath air can be used to reduce the heat haze effect [27, 28]. In this study, we quantified the strain measurement error induced by heat haze using a controlled experiment. Images of hot specimen (923K) please in the test chamber (**Fig 1b**) were taken with and without sheath air. It was found that the strain error, for both cases, is in the range from 0 to 200 $\mu\epsilon$, which is in the typical strain error of DIC at nearly ideal test conditions [28]. It is believed that naturally induced upward convective air flow outside the small test chamber effectively served as the sheath air which reduced the heat haze effect. Hence, in the following experiments, sheath air was not applied. The details of the DIC measurement error analysis regarding the effect of sheath air are given in the **appendix 1**.

Acoustic emission (AE) is an effective non-destructive method for examining the deformation of materials under stress and has been used to study damage in ceramic matrix composites [30, 31, 32, 33]. The strong fiber and weak fiber-matrix interface in ceramic composites activate various toughening mechanisms: crack bridging, fiber sliding, and fiber pull out. Hence, the ceramic composite can tolerate significant damage without catastrophic failure, a distinctive difference from the failure behavior of monolithic ceramics. AE is a powerful technique to

monitor the damage initiation and accumulation process in ceramic composites, and has been used to study damage and failure of ceramic matrix composites. Recently, Nozawa et al. [34] studied tensile, compressive, and in-plane shear failure behaviors of nuclear grade CVI SiC_f-SiC_m composite and were able to correlate damage mechanisms via wavelet analysis of acquired AE signals. They found that due to strong frictional stress at the fiber/matrix interface, AE signals corresponding to matrix cracking occurred prior to the proportional limit stress and thus they were proposed a new failure envelope using the Tsai-Wu criterion [35]. To enable the monitoring of AE in this study, the composite specimen was mounted on a long surrogate tube which extends out of the hot zone. An AE sensor was attached to the cool end of ceramic surrogate tube which functions as a wave guide with low attenuation. The timing, energy, and frequency of AE events can be used to analyze damage progression in the ceramic composites. Together with strain data, the AE signals allow us to detect the threshold strain level when matrix cracking initially occurs.

Finite element method (FEM) is a powerful numerical simulation technique that can be used to solve complex initial- and boundary-value problems. To help understand the results of the experiment, finite element models were constructed to simulate the complex state of temperature, and stress/strain distribution in the test assembly during thermomechanical testing. In the FEM models, all materials are assumed to be linear elastic. While this is a typical assumption that has been used successfully for monolithic materials, it should be noted that this is only a valid assumption for the initial linear elastic region of the SiC_f-SiC_m composite. Temperature was modeled by applying a heat flux at the inner surface, conduction through the bulk. Radiation and convection boundary conditions are applied on the outer surface of the test assembly. The applied heat flux is estimated by the known electrical power consumed by the Nernst glower.

Once the heat flux is imposed, the model calculates temperature distribution in the entire test assembly and predict surface temperature. The obtained temperature field is used to prescribe bulk thermal strain in the decoupled linear elastic stress-strain model. The model is solved using **COMSOL Multiphysics 5.0**, a commercial finite element modeling and simulation software. Most materials properties data needed in the model were obtained from literature [36,37,38], as summarized in **Table 1**. Temperature-dependent thermal conductivity and thermal expansion coefficient were used in the model. The boundary conditions and governing equations used in our FEM models are given in **Table 2**. We have also conducted our own measurements to determine several key material properties specific to our experimental design. In particular, a high temperature experiment using an alumina surrogate and SiC composite was conducted to estimate the thermal resistance of the thin ceramic bond layer in between the SiC sample and surrogate tube. The details are provided in **appendix 2**.

3.1 Experimental

The method of sample preparation for testing is detailed in the next paragraphs. The monolithic SiC sample and SiC_f/SiC_m composite sample were first cut into appropriate lengths using diamond cutting medium. Alumina surrogate tubes were trimmed by turning (using a single point diamond tool) and grinding to make it fit into the SiC sample without damaging the inner surface. The outside diameter of the surrogate was reduced in 0.12 mm for the SiC_f/SiC_m composite sample, and 0.01 mm for the monolithic SiC sample. The size of the sample and surrogate are summarized in **Table 3**. Before inserting the surrogate into the specimen, a thin layer of ceramic adhesive (Ceramabond 503, Aremco Inc) was applied to the outer surface of the surrogate to create a bonding layer between the tubes. **The ceramic adhesive interfacial bonding layer was dried but not sintered before the test.** A bonded SiC_f-SiC_m composite sample and

alumina surrogate tube assembly is shown in **Fig. 3**. A fine wire B-type thermocouple was bonded at the inner surface of the surrogate tube, a K-type fine wire TC was embedded in the ceramic adhesive bond layer between the surrogate tube and the SiC specimen, and another K-type fine wire TC was bonded on the outer surface of the SiC sample. **Because the diameter of the fine wire K-type TC is bigger than the bond layer thickness, a thin groove was ground on the outer surface of the surrogate tube to accommodate the TC.** Temperature data were recorded using a NI cDAQ-9172 chassis with modules NI cRIO-9215 and 9211, and a LabView program.

Figure 3.

Following sample bonding, ZYP coating was applied on the outer surface of SiC sample to create a random speckle contrast pattern to enable DIC measurement. To make a uniform speckle pattern, the ZYP ink was sprayed horizontally, the ink plume was positioned approximate 0.6 m above the sample, and the spray droplets drifting downwards slowly coated the sample surface. After spray coating, the sample was air dried for 12 hours.

The arrangement of the optical components for 3D DIC is shown in **Fig. 4**. Two CCD cameras from Point Grey (GRAS-30S5M) were used for image acquisition; the CCD sensor in the camera is ICX625AL from Sony, and two high resolution fixed focal length (0.025m) lenses (NT63-780, from Edmund Optics). The camera was placed at 0.22 m away from the sample and the angle between the two optical axes is ~ 10 degrees. Two Thorlabs-LIU004 LED lights were used to illuminate the sample which was placed behind a fused quartz window having a transmittance of > 0.9 in the range from 280 nm to 2000 nm. Two 450 nm bandpass filters from Edmund Optics (BP450) were used in front of each camera to filter the light reaching the camera lenses. 450nm is selected so that the wavelength stays within the sensitive spectra window of the

Sony CCD sensor and sufficiently away from thermally emitted light at longer wavelength. During testing, the two CCDs were triggered simultaneously at 4 Hz.

Figure 4.

The AE equipment used in this work is a Micro-II Digital AE System (Physical Acoustics Corporation) equipped with a NANO-30 AE sensor and 60 dB pre-amplifier and the data is analyzed using AEWin software. The AE sensor is attached onto the surrogate tube at a distance of ~ 0.12 m from the specimen. A rubber band was used to hold the AE sensor in place, Vaseline was used as couplant between the sensor and the surrogate tube. A fan was used to air cool the surrogate tube extending out of the hot zone. The sensor temperature was maintained at < 434 K. During test, the AE signal trigger threshold was set at 27 decibels, the sampling rate was set at 5 million sample/sec with analog signal pre-amplified by 60 dB.

To start the test, the YSZ rod was pre-heated to ~ 874 K using an electrical resistance heater (NiCr), and then a voltage of ~ 100 VAC was applied. A few seconds before the rod was moved into final position, the image acquisition, the temperature data acquisition and the AE data acquisition programs were launched; the data were synchronized using recorded time stamp from a network time server. After glower insertion, the specimen/surrogate pair heated up rapidly from the inside out, in a transient heat transfer process which includes intense radiative, conductive, and convective heating until thermal equilibrium was reached, typically over the course of 90 seconds. The applied voltage was then increased at a rate of 0.04 V/s in order to keep a controlled increase of temperature up to the desired temperature as measured by thermocouples located at the I.D. and O.D. of the test specimen. The test was ended after the

target temperature was reached or the maximum current of 4.5 Amp was reached in the electrical circuit.

After testing, the samples were analyzed by optical microscopy and X-ray tomography. A long-working distance optical microscope LEICA CTR6000 with magnification of 0.35x, was used to observe the surface features, and an X-radia MicroXCT 400 X-ray tomography system was used to analyze the internal structure of the sample. The X-radia system is equipped with a 50kV source and Macro-70 magnification (optical magnification 0.39).

4 Results and Discussion

4.1 SiC_f-SiC_m Composite Specimen

Figure 5 shows the representative temperature and strain history for SiC_f-SiC_m composite specimen (type A). The test was conducted up to ~1350 K as measured on the outer surface of the specimen. T_{out} and T_{middle} are the temperature readings of the TCs bonded on the outer surface and in the ceramic bond layer, respectively. The temperature difference ($T_{middle} - T_{out}$) increases as the power of the heater increases reaching a maximum of 250 K when T_{out} is 1350 K. Despite its small thickness, the adhesive bond layer is porous. This will lead to a finite thermal contact resistance [39] between the surrogate and the specimen. The temperature at the outer surface of the surrogate is believed much higher than the temperature of the inner surface of the specimen. Due to its finite size, the TC in the bond layer senses an average temperature in a region across the surrogate tube and the bonding layer. Also, the TC in the bond layer is partially embedded in a groove cut in the surrogate tube, hence its reading will be biased towards the hotter surrogate tube. As such, the actual temperature difference across the SiC_f-SiC_m sample is believed to be less than ($T_{middle} - T_{out}$). The sudden rise in temperature at 117 seconds

corresponds to the time when the heater was inserted into the position inside the surrogate. At ~215 seconds T_{out} was increased at the rate of 0.45 K/s by increasing the voltage in the circuit.

Figure 5.

Figure 6 are representative surface strain maps captured on the outer surface of the SiC_f - SiC_m composite specimen (specimen A) at ~1350 K using the DIC method. The strain distribution on the surface is quite non-uniform. It is believed to be a result of surface texture of the composite. Some of the color bands in the strain map can be seen to mimic the braid pattern of fiber tow. The DIC strain indicated near the border of mapped region is not as accurate as the strain in the central region of the map, because of local high image correlation error due to the curvature of the sample.

Figure 6.

The total strains in **Fig 5** are derived from the DIC measurement. The reported strain value is the average value of a 5 mm x 10 mm window in the center of the specimen. The total strain is resolved into two components, the total axial (ϵ_a) strain and the total hoop strain (ϵ_h). After subtracting the thermal strain (due to temperature rise), the mechanical strain (due to mechanical stress) can be calculated. The thermal expansion coefficient of nuclear grade SiC_f - SiC_m as a function of temperature from Snead et al [38] is used to calculate the thermal strain. Clearly, the strain-time profiles follow the temperature-time profile as expected. From 117 to 200 seconds ($T_{out}=1130$ K) the specimen and surrogate are in a state of transient heat transfer process (thermal shock) with high temperature gradients through the walls of specimen and surrogate, during this period, rapid rise of ϵ_{ma} and ϵ_{mh} were observed. After 200 seconds, the temperature gradient stabilized. The sample-surrogate assembly is in a quasi-steady state. To increase T_{out} beyond

1130 K (200 s), the heater power was slowly ramped up to maintain a rate of ~ 0.45 K/s in T_{out} till the end of the test.

At the beginning of the transient phase, the mechanical hoop strain (ϵ_{mh}) is higher than the mechanical axial strain (ϵ_{ma}) up to 1050 K. From this point forward, the mechanical axial strain (ϵ_{ma}) becomes larger than the mechanical hoop strains (ϵ_{mh}) till the end of the test. It is believed that relaxation of ϵ_{mh} has occurred due to cracks developed along the axial directions of the sample. The sample is in a state of biaxial strain with a varying strain ratio ($\epsilon_{\text{ma}}/\epsilon_{\text{mh}}$), which starts at zero at 149 seconds, rises to 1.0 at 175 seconds (1077 K), and then slowly increases to 1.1. The error of the total DIC strains is around $250 \mu\epsilon$ during the period of test.

The maximum combined thermal and mechanical strain measured during testing reaches approximately $6000 \mu\epsilon$ in both the axial and hoop direction. This corresponds to mechanical strain of approximately $1300 \mu\epsilon$. Based on room temperature stress versus strain data obtained by burst test similar to the test shown in **Fig. 2**, this corresponds to approximately 150 MPa hoop stress at the OD of the composite and 220 MPa at the ID of the composite. This stress/strain values are well past the linear elastic region of the composite, but short of the UTS which is typically a minimum of 200 MPa and $3000 \mu\epsilon$. It should be noted that this tensile data was collected at room temperature, however, for CVD based $\text{SiC}_r\text{-SiC}_m$ the shape and magnitude of stress/strain plots tends to undergo minimum changes up to temperatures of 1500 K [38].

The captured AE signals provide additional clues to the observed strain temperature history. In **Figure 7**, T_{out} , ϵ_{ma} , ϵ_{mh} , and cumulative AE energy are plotted vs. time. The cumulative AE energy starts to increase shortly after the heater was inserted in position. The largest activity of AE signals occurs during the transient phase and the beginning of the constant temperature rate

phase up to 349 seconds, reaching to an accumulative AE energy of 7.1E^{-12} J at the end of the test. Above 1286 K the cumulative AE energy plateaus, this indicates the termination of any cracking events.

Figure 7.

The individual AE events (and their energy) are shown in **Fig. 8** where the discrete bursts of energy can be better observed. It is believed that the main source of AE activity is originated from the SiC specimen. It is unlikely from the surrogate tube, which is in a state of compressive stress and shows no damage in post-test examination. It is also unlikely that significant acoustic events are emanating from the ceramic bond layer. Before testing, the ceramic bonding layer is in a dried green compact state, i.e., aggregates of alumina powders with binders and pores. Under compressive radial stress, the layer is expected to further compact and gradually acquire strength as the small particles begin to sinter together during the high temperature excursion. Post-test examination showed no visible damage in the surrogate tube and interfacial bonding layer by optical microscopy and X-ray tomography. These facts, coupled with FEM analysis, described below, suggests that the SiC/SiC tube is the main source of high energy AE events.

Figure 8.

The first high energy AE burst ($\sim 4.4\text{E}^{-13}$ J) occurred when the surface mechanical hoop strain reaches ~ 250 $\mu\epsilon$ registered in the center of the sample tube. Based on the model results described below, the hoop strain at the inner surface of the sample near the cut edge of the tube is significantly higher than the average surface hoop strain in the center of the tube. There, the material has very likely reached the PLS. A second high energy burst of 8.4E^{-13} J occurs at the end of the transient phase, corresponding to a hoop strain of ~ 700 $\mu\epsilon$. After the surface hoop

strain reaches 1100 $\mu\epsilon$, the AE burst stops. It is believed that the SiC_r-SiC_m composite sample has traversed PLS in this period, and acoustic emissions events are a result of further matrix micro-cracking and/or fiber sliding/fracture. After damage has been accumulated in the composite, it becomes more compliant, and the solid surrogate tube become ineffective of exerting more internal force to the sample due to its finite expansion. AE burst events thus are terminated. The tested sample still retains a high level of mechanical integrity. After removing from the surrogate tube, the post-test SiC_r-SiC_m samples are indistinguishable by their appearances from untested material. Careful inspection under optical microscope (LEICA CTR6000, 50x magnification) revealed microcracks in the inner surface of the sample, as shown in **Fig. 9**, confirming the specimens was damaged beyond the PLS and that maximum stress is at the ID of the composite

Figure 9.

In order to gauge the effect of test specimen's length on the test results, two additional lengths of SiC_r-SiC_m composite specimens (type B and C) were tested. The length of sample type B is 8.5 mm, and the length of sample type C is 4 mm. Similar sample preparation and testing procedures were followed. In **Figure 10**, the temperature and strain history of specimen B are plotted. The total strains follow the temperature profile, as expected. The axial (ϵ_{ma}) and hoop (ϵ_{mh}) components of the mechanical strains are equal in magnitudes at the beginning of the test; however, the axial mechanical strain become larger than the hoop mechanical strain after the transient phase till the end of the test. The same temperature ramp rate of ~ 0.45 K/s was applied after then transient phase of the test. The strain ratio, $\epsilon_{ma}/\epsilon_{mh}$, grows from 0 to 1.03 (273 to 832 K), then slightly decreases down to 0.97 and then continues to increase until the end of the test reaching a maximum value of 2.7. This indicates that the micro-cracking has relieved most of

the hoop stress, the sample is still bonded to the surrogate and thus being forced to extend in the axial direction with the surrogate tube. The DIC error is around 250 $\mu\epsilon$, increasing at the end of the test to 800 $\mu\epsilon$. This is probably caused by micro cracking at the surface and the spallation of the speckle pattern.

Figure 10.

In **Figure 11**, the mechanical strains and the cumulative AE energy are plotted vs. time for sample type B. As in the case of specimen A, the largest activity of AE signals occurs during the transient phase and the beginning of the constant temperature ramp phase up to 280 seconds, reaching a value of 4.2E^{-12} J at the end of the test. Above 1300 K the AE energy increases slightly. The individual AE events are shown in **Fig. 12** where bursts of energy are observed at difference times during the test. The first high energy event (4.1E^{-13} J) occurred when the surface mechanical hoop strain reached $\sim 731 \mu\epsilon$, and an energy emission of 3.5E^{-13} J is observed close to the end of the test. During post-test examination under a microscope, a single thin surface crack was observed orientated along the axis of the sample tube.

Figure 11.

Figure 12.

Similar results were observed for all types of composite specimen in terms of temperature, strain and AE response. No macroscopic cracks can be seen by unaided visual examination for all specimen types. Micro-cracking at the ID was observed for specimen type A and micro cracking at the OD were observed for specimen type B and C. **Figure 13** shows the X-ray tomography image of a composite specimen (type C) after test with maximum $T_{\text{out}} = 1351$ K, showing a micro crack running through the thickness of the specimen. The through-crack

appears to have been deflected multiple times when advancing through composite wall. Despite the through thickness crack it does appear that some ability to carry stress and strain is retained in the specimen as no sudden drops in strain were observed during testing.

Figure 13.

An interesting pattern of surface mechanical strain history is observed for all type of SiC-SiC samples. Initially, the measured axial strain on the sample surface is close to the measured hoop strain. This is consistent with the expected bi-axial stress state in which roughly equal amounts of stress are observed in both the axial and hoop direction, as demonstrated in the FEM described below. After a period of intense AE events, the axial strain starts to exceed hoop strain as the test progresses. Together with post-examination evidences, it is believed that microcracking is occurring during the thermos-transient phase (evidenced by the intense AE events), and these cracks runs primarily along the axial direction as seen in the micrograph. Such cracks relax the hoop stress more than the axial stress, and hence results in lower hoop strain than the axial strain.

This effect seems to be more pronounced as size of the specimen is decreased. The type A specimens were overcoated with SiC at General Atomics so that the cut ends were sealed and protected from the environment. However, for type B and C composites this overcoating was not performed. It is believed that in type B and C specimens, oxidation of the PyC interphase coating occurs which leads to a more brittle like composite behavior and additional through thickness cracking takes place. The FEM model described below reveals a pronounced edge affected zone where the hoop stress near the free edge of the tube segment is significantly higher than the hoop stress in the center of the tube. For shorter specimen, the edge affected zone represents a much larger fraction of the samples. It is believed that more cracking (along the axial direction) occurred in shorter specimen. This resulted in more pronounced relaxation of

hoop strain. Additional testing is currently being performed to validate this theory, which may underscore the need to select sample of adequate length and properly protect the interphase layer from degradation during exposure to severe conditions.

4.2 SiC_f-SiC_m Composite Modeling

FEM models were constructed to predict temperature, stress, and strain distribution through the thickness direction and length of the test assembly. In order to simplify the modeling, the composite material is assumed to be linear-elastic orthotropic material. As such, the FEM model will not be able to capture the pseudo-plastic behavior occurs beyond the PLS of SiC_f-SiC_m composite. The thermo-mechanical properties of surrogate, bond layer, and specimen used in the model are shown in Table 1. The constitutive equations solved in the model are shown in Table 2. The SiC_f-SiC_m sample is modeled as a four-layer orthotropic composite with fiber orientation +45°/-45°/+45°/-45° on each layer. The dimensions of surrogate, bond layer and specimen are given in Table 3. The modeling domain and the applied boundary conditions are illustrated in **Figure 14**. The model solves a heat transfer problem where radiation, convection and conduction of heat are involved, having as boundary conditions the ambient temperature and a heat flux on the inner surface of the surrogate that extends in axial direction in a length similar to the specimen's length. Heat conduction occurs through the material, and convection and radiation occurs on the outer surfaces.

Figure 14 also shows the mesh used in the FEM model and the predicted deformed shape (exaggerated) and surface temperature distribution at the moment when the outer surface mechanical strain reaches about 600 $\mu\epsilon$. The deformed SiC_f-SiC_m sample shows a characteristic wheat sheath (or hourglass) shape due to the transition from plane strain (near the center) to

plane stress (at free end) state. The fuel pellet shows similar deformed shape when temperature gradient exists [40]. **Figures 15~17** show the through-thickness distribution of temperature, stress and strain near the center of the specimen. Due to the porous nature of ceramic paste, its thermal conductivity is much lower than dense ceramic material of the same composition. A large temperature drop across the ceramic paste is predicted. The model predicts that the surrogate is in compression, while the SiC composite sample and the bond layer are in tension. The predicted mechanical strain is slightly higher than the measured mechanical strain at the corresponding outer surface temperature ($T_{out}=1087K$). The axial mechanical strain distributed through the wall is very uniform, while a gradient of hoop mechanical stress is predicted. **The highest hoop stress and strain occurs at the inside surface of the SiC composite sample. This is also the location where the hoop cracks were observed.**

Figure. 14

Figure 15.

Figure 16.

Figure 17.

Figure 18 shows the stress and strain distribution along the length of the sample. Much higher hoop strain can be found near the free edge of the sample due to the transition from plane strain to plane stress conditions near the edge. In comparison, the axial stress/strain build up from the free edge to the center of the sample due the shear lag, as expected (**Fig. 1c**). Due to the high hoop stress/strain in the edge affected zone, shorter the sample will experience high hoop stress in a larger volume. This is believed to be one of the root causes for the different responses observed in sample type A, B, and C.

Figure 18.

4.3 Monolithic SiC Specimen

The monolithic SiC specimen was tested following the same procedure as composite samples. The temperature and strain vs. time are shown in **Fig. 19**. After the insertion of the heater (47 seconds) there is a slight decrease in mechanical strains indicated by DIC. It is believed to be due to a slight movement of the sample due to a small impact during glower rod insertion. A peak of temperature and strain are observed in the transient phase right after the glower insertion. This is attributed to a brief glower power excursion due to the slow response of the ballast. During the thermal transient, the mechanical hoop strains increased sharply, much more than the mechanical axial strains. After the transient phase the sample was slowly heated at a rate of ~ 0.45 K/s (T_{out}) till the end of the test as in the previous tests. After the peak, the mechanical hoop strain decreases to almost zero and then start to increase again. The mechanical axial strains decrease in a small amount after the peak of power, continuing to increase until the end of the test. The mechanical strain ratio ($\epsilon_{ma}/\epsilon_{mh}$) varies strongly during the test with mechanical hoop strains dominant during the transient phase and the mechanical axial strains dominant during the constant temperature rate phase. The DIC error is around $250 \mu\epsilon$, increasing at the end of the test to $400 \mu\epsilon$.

Figure 19.

In **Figures 20** and **21**, AE signals are combined with temperature and strain history for the monolithic SiC specimen. As observed from the graph, the most AE activities occur during the transient phase, and the highest emission of energy corresponds to the time at the peak power of the heater. Compared to SiC_f - SiC_m composite samples, the AE events in monolithic sample are

rather sparse, but of much higher energy. The more-brittle response is expected due to the lack of crack bridging and deflection mechanisms in the monolithic SiC samples. Despite strong AE evidence of sample cracking, no macroscopic cracks were seen on the outer surface of monolithic SiC sample. The cracks in the SiC monolith sample are believed to initiate near the inner surface of the SiC sample tube during thermal transient phase, due to high hoop strain; the hoop cracks propagate towards outer surface and into a stress field of lower intensity. Once the hoop cracks forms near the inner surface, the sample tube become more compliant, this further reduce the hoop stress in outer surface. Unlike gas or liquid pressurization method, the solid surrogate tube has much lower total stored strain energy, the crack driving force subsides quickly after the sample become more compliant. It is believed that the combined effect of the above factors result in the observed crack stabilization in the monolithic SiC sample. In comparison, the same SiC monolithic sample tube pressure tested in the authors' lab using polymer and liquid pressurization media at room temperature had all shattered explosively at the point of failure.

Figure 20.

Figure 21.

4.4 Monolithic SiC Modeling

Model predicted temperature profile, stress and mechanical strain distribution through the thickness of SiC monolithic sample is shown in **Fig. 22** and **23**. The predicted axial stress is slightly smaller than the axial stress at the outer surface. The strain profile is similar to that of composite sample, see **Fig. 23**. Similar edge affected zones exist for monolithic samples. The predicted hoop mechanical strain on the outer surface is somewhat smaller than actually measured hoop mechanical strain at the corresponding outer surface temperature ($T_{\text{out}} = 881 \text{ K}$),

and the predicted axial mechanical strain which is higher than the measured mechanical axial strain. Based on the prediction, the most vulnerable location for micro cracks to initiate is at the inner wall of the SiC sample.

Figure 22.

Figure 23

5 Conclusions

Compared to other high temperature burst test methods, the novel test scheme used in this study is much simpler, safer and more affordable. It appears to come close to simulating LOCA accident condition with high temperature gradient and direct mechanical interaction by using an expanding solid surrogate. A high temperature image acquisition method using a filtering technique enabled the application of 3D DIC method to capture the strain distribution on the outer surface of the specimen. AE signals are found indispensable for understanding dynamic events occurring in the sample during testing. Finite element modeling confirms the state of biaxial stress as observed in experiment and shed lights on the stress distribution across the thickness direction of the sample. Further optimization of the test method and applications to SiC_f-SiC_m samples of different property, overcoated length and architectures are currently underway.

The mechanical response of nuclear grade SiC_f-SiC_m during testing showed encouraging results. The evidence suggests that the composite sample suffered matrix microcracking when its surface mechanical hoop strain reached $\sim 300 \mu\epsilon$, however the additional fracture toughness provided by the fiber reinforcement meant the specimen retained further load bearing ability. Material relaxation occurred due to this micro-cracking which weakened the interaction with the

solid surrogate. The sample was found in a biaxial state of stress, the hoop component peaked and relaxed in a more significant manner than the axial component. Based off AE data, the burst of damage events primarily occurs in a short high temperature transient window suggesting that both mechanical stress and thermal shock is playing a role in damage evolution. The monolithic sample showed similar but more brittle behavior, as expected. The composite sample retained a high level of mechanical integrity in the simulated LOCA conditions used in this test, suggesting an engineered SiC/SiC based composite cladding would show an excellent ability to retain fuel in accident conditions. During LOCA conditions, localized micro damage may occur in the engineered SiC/SiC based composite cladding due to thermal and mechanical stress, particularly during rapid transient events. However, based on this test data, as the composite enters the post PLS region of the stress strain curve, damage evolution stabilizes and the composite enters a pseudo stable state where additional thermal and mechanical load does not cause further damage. Under similar temperatures as those used in this test, metals based materials such as Zircaloy or FeCrAl would have lost more than 90% of their room temperature strength [41] and would be susceptible to ballooning and/or bursting. These results, coupled with the low rates of hydrogen production as demonstrated in the literature [42] suggest SiC/SiC as an excellent candidate for accident tolerant fuel.

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Appendix 1. Quantification of heat haze effect on DIC strain measurement

Heated SiC sample tube was placed on a translation stage. DIC cameras were setup to image the sample through a viewing window before and after a small rigid body translation was applied to the sample. The tests were conducted with and without a sheath air was applied to the viewing window. Results of the study are shown in table 4. Based in the results, the axial and hoop strains measured with and without air sheath are comparable, considering the typical measurement error of DIC techniques is $200\ \mu\epsilon$. The difference of about $130\ \mu\epsilon$ in the hoop strain measured with and without air sheath, is probable due to heat haze occurring outside the quartz window used to acquire the images.

Appendix 2. Estimation of the thermal conductivity of the ceramic adhesive bond layer

It was found through parametric study that thermal conductivity of the bond layer has great influence on the temperature distribution. Hence, we conducted a series of experiments to help estimate the thermal resistance of the bond layer. A known heating power (via a centrally located Ni-Cr resistance heater) is applied through SiC sample, alumina surrogate, and bonded tube assemblies in separate experiment. Temperature gradients were measured across the thickness of the sample. The results are shown in the Table 5. The conductivity of the bond layer was calculated using the thermal resistance of the bond layer:

$$k_I = \frac{t_I}{R_I A_I}$$

Where t_I is the thickness of the bond layer, R_I is the thermal resistance of the bond layer, and A_I is the area of the interphase. The thermal-contact resistance of the bond layer is calculated using the equation:

$$R_I = R_T - R_A - R_{SiC} - R_{Contact}$$

R_T , R_A , R_{SiC} , and $R_{contact}$ are the total thermal resistance, thermal resistance of alumina, SiC, and parasitic contact resistance obtained from separate experiments. The bond layer thermal conductivity was found to be ~0.55 W/mK.

Figure and Table Captions:

Figure 1. (a) Schematic of experiment, (b) the test chamber, and (c) stresses acting on the tubing

Figure 2. DIC vs. strain gage stress-strain curves of nuclear grade SiC_f-SiC_m composite obtained by burst test at room temperature.

Figure 3. Picture of SiC_f-SiC_m composite sample mounted on surrogate tube

Figure 4. Schematic of the optical imaging setup for 3D DIC measurement.

Figure 5. Temperature profiles and strains for composite specimen type A

Figure 6. 3D-strain map at 1350 K of composite specimen type A from DIC measurement, (a) total axial strains, (b) total hoop strains

Figure 7. Mechanical strains and cumulative AE energy for composite specimen type A

Figure 8. Mechanical strains and individual AE events for composite specimen type A

Figure 9. 50x magnification microscope images of inner surface of sample A

Figure 10. Temperature profiles and strains for specimen type B

Figure 11. Mechanical strains and cumulative AE energy for specimen type B

Figure 12. Mechanical strains and individual AE events for specimen type B

Figure 13. X-ray images of a type-A SiC composite specimen after test.

Figure 14. FEM modeling domain and boundary conditions (left); FEM mesh and representative deformed shape and surface temperature distribution of the sample (right).

Figure 15. Model predicted temperature profile and stress distribution through the wall at $T_{out} = 1087K$ for SiC_f-SiC_m sample.

Figure 16. Model predicted temperature profile and total strain distribution through the wall at $T_{out} = 1087K$ for SiC_f-SiC_m sample

Figure 17. Model predicted temperature profile and mechanical strain distribution through the wall at $T_{out} = 1087K$ for SiC_f-SiC_m sample

Figure 18. Model predicted mechanical stress distribution along the length of the sample at $T_{out} = 1087K$ for SiC_f-SiC_m sample type A

Figure 19. Temperature profiles and strains for monolithic SiC

Figure 20. Plot of mechanical strains and cumulative AE energy for monolithic SiC

Figure 21. Plot of mechanical strains and AE energy for monolithic SiC

Figure 22. Model predicted temperature profile and stress distribution through the wall at $T_{OUT} = 881 K$ for the monolithic SiC

Figure 23. Model predicted temperature profile and mechanical strain distribution through the wall at $T_{OUT} = 881 K$ for the monolithic SiC

Table 1: Material properties used in FEM simulation

Table 2: Constitutive equation and boundary conditions used in the FEM model

Table 3: Dimension of SiC samples and surrogate tubes

Table 4. DIC strain in the controlled displacement experiment.

Table 5. Experimental data used to estimate the thermal conductivity of the interphase

Figure 1. (a) Schematic of experiment, (b) the test chamber, and (c) stresses acting on the tubing

Figure 2. DIC vs. strain gage stress-strain curves of nuclear grade $\text{SiC}_f\text{-SiC}_m$ composite obtained by burst test at room temperature.

Figure 3. Picture of SiC_f-SiC_m composite sample mounted on surrogate tube

Figure 4. Schematic of the optical imaging setup for 3D DIC measurement.

Figure 5. Temperature profiles and strains for composite specimen type A

Figure 6. 3D-strain map at 1350 K of composite specimen type A from DIC measurement, (a) total axial strains, (b) total hoop strains

Figure 7. Mechanical strains and cumulative AE energy for composite specimen type A

Figure 8. Mechanical strains and individual AE events for composite specimen type A

Figure 9. 50x magnification microscope images of inner surface of sample A

Figure 10. Temperature profiles and strains for specimen type B

Figure 11. Mechanical strains and cumulative AE energy for specimen type B

Figure 12. Mechanical strains and individual AE events for specimen type B

Fig. 13. X-ray image of a type A SiC composite specimen after test.

Figure 14. FEM modeling domain and boundary conditions (left); FEM mesh and representative deformed shape and surface temperature distribution of the sample (right).

Figure 15. Model predicted temperature profile and stress distribution through the wall at $T_{\text{out}} = 1087\text{K}$ for $\text{SiC}_r\text{-SiC}_m$ sample.

Figure 16. Model predicted temperature profile and total strain distribution through the wall at $T_{\text{out}} = 1087\text{K}$ for SiC_r-SiC_m sample

Fig. 17. Model predicted temperature profile and mechanical strain distribution through the wall at $T_{\text{out}} = 1087\text{K}$ for $\text{SiC}_r\text{-SiC}_m$ sample

Fig. 18. Model predicted mechanical stress distribution along the length of the sample at $T_{\text{out}} = 1087\text{K}$ for $\text{SiC}_f\text{-SiC}_m$ sample type A

Figure 19. Temperature profiles and strains for monolithic SiC

Figure 20. Plot of mechanical strains and cumulative AE energy for monolithic SiC

Figure 21. Plot of mechanical strains and AE energy for monolithic SiC

Figure 22. Model predicted temperature profile and stress distribution through the wall at $T_{OUT} = 881$ K for the monolithic SiC

Figure 23. Model predicted temperature profile and mechanical strain distribution through the wall at $T_{OUT} = 881$ K for the monolithic SiC

Table 1: Material properties used in FEM simulation [35,36,37,40]

	SiC Composite	Alumina Surrogate ⁽⁴⁾	Bond Layer ⁽⁵⁾	Monolithic SiC
Elasticity Modulus, E (GPa)	E ⁽¹⁾	359	18	332 ⁽⁶⁾
CTE, α (1/K)	α ⁽²⁾	α ⁽⁴⁾	3E-6	α ⁽⁷⁾
Thermal Conductivity, k (W/mK)	k ⁽³⁾	k ⁽⁴⁾	5	k ⁽⁸⁾
Poisson Ratio, ν	0.2 ⁽¹⁾	0.25	0.1	0.17 ⁽⁹⁾

⁽¹⁾Anisotropic material properties, Kuo et al [40].

$$E_{11} = 3.2\text{E}11 \text{ GPa}; E_{22}=E_{33}= 2.86\text{E}11; \nu_{12} = \nu_{23} = 0.2; G_{12} = G_{13}=9.5\text{E}10$$

⁽²⁾from Snead et al., Journal of Nuclear Materials 371 (2007) 329-377.

$$\alpha = -1.8276 + 0.0178 T - 1.554 * 10^{-5} T^2 + 4.5246 * 10^{-9} T^3 \quad (10^{-6}/\text{K}), T \text{ in Kelvin}$$

⁽³⁾ Temperature dependent data provided by General Atomic..

$$k_p = [-0.0003 + 1.05 * 10^{-5} T]^{-1} \quad (\text{Wm}^{-1}\text{K}^{-1}), T \text{ in Kelvin}$$

⁽⁴⁾National Institute of Standards and Technology (NIST) at www.ceramics.nist.gov

⁽⁵⁾ Properties were assumed.

⁽⁶⁾Measured by impulse excitation method in house.

⁽⁷⁾ Measured by dilatometry in-house.

⁽⁸⁾Temperature dependent from R.G. Munro, “Material properties of a sintered α -SiC”, Ceramics Division, NIST, 1997.

$$k = \frac{52000 e^{-1.24 \cdot 10^{-5} T}}{T+437} \quad (\text{Wm}^{-1}\text{K}^{-1}), T \text{ in Kelvin.}$$

⁽⁹⁾from Ortech Inc.

Table 2: Constitutive equations and boundary conditions used in the FEM model

Constitutive Equations (Hook's law, for isotropic material in the model)	Radiation (W/m ²)	Convection (W/m ²)	Conduction (W/m ²)
$\varepsilon_r = \frac{1}{E}(\sigma_r - \nu(\sigma_\theta + \sigma_z)) + \alpha(T - T_r)$ $\varepsilon_\theta = \frac{1}{E}(\sigma_\theta - \nu(\sigma_r + \sigma_z)) + \alpha(T - T_r)$ $\varepsilon_z = \frac{1}{E}(\sigma_z - \nu(\sigma_\theta + \sigma_r)) + \alpha(T - T_r)$	$q_{rad} = \epsilon \sigma_B (T_a^4 - T)$	$q_{conv} = h_a(T_{ext} - T)$	$q_{cond} = -k \Delta T$
Boundary conditions: <ul style="list-style-type: none"> - Heat flux on inner surface of surrogate, 360 kW/m² - Ambient temperature, T_a = 294 K 			

Table 3: Dimension of SiC samples and surrogate tubes

Parameter	Monolith SiC	SiC Composite	Surrogate
Outer Diam. O.D. (mm)	9.5	10.63	6.35 and 7.95
Thickness (mm)	1.58	1.4	0.79 and 1.2
Length (mm)	9.0	12.7 (type A) 8.5 (type B) 4.0 (type C)	10 and 140

Table 4. DIC strain in the controlled displacement experiment.

Sheath Air	Axial Strain ($\mu\epsilon$)	Hoop Strain ($\mu\epsilon$)
with	68	61
without	93	193

Table 5. Experimental data used to estimate the thermal conductivity of the bond layer

Heat Flux (W)	T_{in} (K)	T_{mid} (K)	T_{out} (K)	R_T (K/W)	k_{SiC} (W/m-K)	R_{SiC} (K/W)	k_A (W/m-K)	R_A (K/W)	$R_{Contact}$ (K/W)	R_I (K/W)	k_I (W/m-K)
1.4	309	301	300	6.4	70.3	0.13	36.0	0.12	5.8	0.31	3.72
2.8	351.7	339	337	5.3	66.9	0.14	31.8	0.14	4.3	0.72	1.60
6.2	528	469	455	11.8	58.0	0.16	19.3	0.23	9.3	2.10	0.55
11.1	672	572	548	11.2	52.4	0.18	13.5	0.33	8.7	1.98	0.58
15.5	785	649	614	11.0	49.1	0.19	10.7	0.41	8.4	2.07	0.55
19.8	891	719	675	10.9	46.4	0.20	9.0	0.49	8.2	2.02	0.57
27.0	1027	813	749	10.3	43.4	0.22	7.7	0.57	7.4	2.15	0.53
32.2	1107	872	797	9.6	41.7	0.22	7.1	0.62	6.7	2.10	0.54
38.3	1212	945	859	9.2	39.7	0.24	6.6	0.67	6.3	2.01	0.57