Design Optimization of Advanced PWR SiC/SiC Fuel Cladding for Enhanced Tolerance of Loss of Coolant Conditions

by
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Abstract

Limited data has been published (especially on experimental work) on integrated multilayer SiC/SiC prototypical fuel cladding. In this work the mechanical performance of three unique architectures of three-layer silicon carbide (SiC) composite cladding is experimentally investigated under conditions associated with the loss of coolant accident (LOCA), and analytically under various conditions. Specifically, this work investigates SiC cladding mechanical performance after exposure to 1,400°C steam for 48 hours and after thermal shock induced by quenching from 1,200°C into either 100°C or 90°C water. Mechanical performance characteristics are thereafter correlated with sample architecture through void characterization and ceramography.

The series with a reduced thickness did not have a pseudo-ductile regime due to overloading of the composite layer. The presence of the axial tow did not yield significant difference in the mechanical behavior most likely because samples were tested in the hoop direction. While as-received and quenched samples behaved similarly (pseudo ductile failure except for one series), non-frangible brittle failure (single-crack failure with no release of debris) was systematically observed after oxidation due to silica buildup in the inner voids of the ceramic matrix composite (CMC) layer. Overall, thermal shock had limited influence on sample mechanical characteristics and oxidation resulted in the formation of silica on the inner wall of the CMC voids leading to the weakening of the monolith matrix and brittle fracture.

Stress field in the cladding design is simulated by finite element analysis under service and shutdown conditions at both the core’s middle height and at the end of the fuel rod. Stresses in the fuel region are driven by the thermal gradient that creates stresses predominantly from irradiation induced swelling. At the endplug, constraints are mainly mechanical. Stress calculations show high sensitivity to the data scatter and especially swelling and thermal conductivity. No cladding with the design studied here can survive either service or shutdown conditions because of the high irradiation-induced tensile stresses that develop in the hot inner monolith layer. It is shown that this peak tensile stress can be alleviated by adjusting the swelling level of the different layers. The addition of an under-swelling material such as PyC or Si can reduce the monolith tensile stress by 10%. With a composite that swells 10% less than the monolith, the stress is reduced by 20%.

Thesis supervisor: Michael Short

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My family was a great constant support even across the Atlantic and I would like to thank them for that.

Milton Cornwall-Brady’s help for gathering the X-Ray CT images at the MIT Animal Imaging and Preclinical Testing Core is gratefully acknowledged.

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List of Figures
Figure 1: SEM cross sectional views of the three layers (left) and two layer (right) designs ..... 13
Figure 2: Optical view of the different WEC Designs .................................................. 16
Figure 3: Comparison CVD and CVI monolith SiC .................................................. 17
Figure 4: As-received SEM views of Tri-Axial-thin Top: cross-section, Bottom: lateral view ... 18
Figure 5: Original (left) XCT and processed (right) image slices of the Tri-Plain series .... 19
Figure 6: Oxidation Facility ..................................................................................... 21
Figure 7: Weight measurement for a 10g/min evaporation flow rate .......................... 22
Figure 8: Thermal shock apparatus (35) .................................................................. 22
Figure 9: X-Ray Image of the three architectures .................................................. 23
Figure 10: Inner pressure vs Displacement for the as-received failure test ................. 24
Figure 11: Failure Pressure vs Thickness for the as-received samples ..................... 25
Figure 12: Failure modes of the as received samples ............................................. 26
Figure 13: Stress-Strain curve for Tri-Axial-Thick .................................................. 28
Figure 14: Failure Probability for Three-layer CMC ............................................... 29
Figure 15: Optical view of sample Tri-Axial-Thick pre and post oxidation ............... 30
Figure 16: Silica formation on the exterior lateral surface after 48 hour oxidation ...... 30
Figure 17: Inner oxidation of the CMC through the voids ........................................ 31
Figure 18: Internal Pressure vs Metal insert displacement for as-received and oxidized samples 32
Figure 19: SEM fracture analysis of as-received and oxidized samples .................... 33
Figure 20: Stress-strain behavior of as-received (blue) and oxidized (red) Tri-Axial-Thick ...... 33
Figure 21: Crack propagation of as-received (left) and oxidized samples (right) ......... 34
Figure 22: SEM view of an exposed fiber pre- (left) and post- (right) quench (series Tri-Plain) 35
Figure 23: Heat transfer regimes during thermal shock for Tri-Plain-Thick .................. 36
Figure 24: Comparison as-received and quenched Inner pressure vs Displacement curves 37
Figure 25: SEM comparison of as-received (left) vs quenched (right) series failure modes .... 38
Figure 26: Stress-strain curve for As-Received (blue), 100°C Quench (red) and 90°C Quench (green) for Tri-Axial-Thick .......................................................... 39
Figure 27: ADINA model, including loads and fixities – not to scale ............................ 40
Figure 28: Modelling end effects on the fuel rod – not to scale ................................. 41
Figure 29: Hoop Stress Decomposition ..................................................................... 45
Figure 30: Hoop Strain, Hoop Stress and Temperature Profiles .................................. 46
Figure 31: Thermal Strain and Hoop Stress: Continuous (Blue), Discretized (Red) Properties . . 46
Figure 32: Comparison of Service (blue) and Shutdown (red) Hoop Stress ................. 47
Figure 33: Stress propagation from fuel to plenum regions ...................................... 49
Figure 34: Swelling Sensitivity Analysis Reference (Red, Equation 17) Upper (Green, Equation 22) Lower (Blue, Equation 21) ................................................................. 50
Figure 35: Hoop stress distribution for the three cases ............................................ 51
Figure 36: Endplug region: Axial Stress for the three endplug designs ....................... 52
Figure 37: Hoop stress distribution over the cladding radius for Case II (a) and Case III (b). 53
Figure 38: Hoop stress at BOL (blue), EOL (orange) and under the CCS (grey) .......... 54
Figure 39: Swelling behavior of PyC (51) ................................................................. 56
Figure 40: Thin layer: 25 µm ................................................................. 57
Figure 41: Stress Profile with an Si Interface ................................................... 58
Figure 42: Swelling Experimental Data and the Monolith & Composite Fittings .......... 59
Figure 43: Stress Results with the under swelling composite ................................ 60
List of Tables

Table 1: Test Matrix: Experimental Investigations ................................................................. 14
Table 2: Test Matrix, Finite Element Analysis ........................................................................ 14
Table 3: Design Parameters for three and two layer specimen ............................................. 16
Table 4: Porosity Results .......................................................................................................... 24
Table 5: As-received strength testing ......................................................................................... 28
Table 6: Weibull parameters for different SiC types ................................................................. 29
Table 7: Oxidation weight change ............................................................................................. 31
Table 8: Comparison of internal pressure at failure for as-received and oxidized samples ...... 32
Table 9: Thermal shock mechanical results .............................................................................. 37
Table 10: Endplug Design Choices ........................................................................................... 41
Table 11: Simulation Conditions for SiC Cladding Stress Calculation ........................................ 42
Table 12: Material properties for FEA analysis ...................................................................... 43
Table 13: Axial Stress extrema on line AB for the different plug designs ............................... 48
Table 14: Thermal conductivity sensitivity study: maximal hoop stress during service and shutdown for different TC value pairs ................................................................. 49
Table 15: PyC & Silicon Properties .......................................................................................... 56
Table 16: Maximum Hoop stress (inner monolith) .................................................................. 57
Table 17: Peak Stress Summary Table ..................................................................................... 61
# Table of Contents

I. Introduction ........................................................................................................... 11
   1. Motivation ........................................................................................................ 11
   2. Objectives and Scope .................................................................................. 13
   3. Description of the specimens .................................................................. 15

II. Experimental work ................................................................................................ 19
   1. X-Ray CT Analysis .................................................................................... 19
   2. Burst Test ..................................................................................................... 19
   3. Oxidation Background & Facility .............................................................. 20
   4. Description of the Thermal Shock Facility ............................................ 22

III. As-Received Results .......................................................................................... 23
   1. Porosity Results .......................................................................................... 23
   2. Mechanical Results .................................................................................... 24
   3. Mechanical Model ...................................................................................... 26
   4. Discussion .................................................................................................... 27

IV. High Temperature Steam Experiment ................................................................. 30
   1. Oxidation Results ....................................................................................... 30
   2. Mechanical Results .................................................................................... 32
   3. Discussion & Summary ............................................................................. 33

V. Thermal Shock Experiments ............................................................................... 35
   1. Thermal Shock Progress ........................................................................... 35
   2. Mechanical Results .................................................................................... 37
   3. Discussion & Summary ............................................................................. 39

VI. Finite Element Analysis under service and shutdown ..................................... 40
   1. Description of the Work ............................................................................ 40
   2. Material Properties ..................................................................................... 43
   3. Results ......................................................................................................... 44
   4. Sensitivity Analysis .................................................................................... 49
   5. Summary .................................................................................................... 54

VII. Tensile Stress minimization of the three-layer design .................................. 55
   1. Introduction .................................................................................................. 55
   2. The Thin Layer Option .............................................................................. 56
I. Introduction

1. Motivation

Nuclear fuel cladding

Nuclear fuel cladding has several key functions to ensure for a nuclear power plan to operate safely. The cladding is the first barrier to radioactivity release and should contain fission products and gases under both service and accident conditions. It should also ensure the heat transfer from the fuel to the coolant while maintaining safety margins. The challenging environment of a nuclear core complicates the material's selection process because the cladding has to support radiation damage, corrosion, several types of degradations (grit-to-rod fretting or flow induced vibrations) and mechanical loads (pressurization, thermal stresses, and swell-induced stresses) and should have a low neutronic penalty. As such, it must meet strict thermomechanical and radiation performance criteria including maintaining strength at high temperatures and high radiation fields with low swelling and neutron absorption. Those requirements become more critical under accidental conditions such as a loss-of-coolant-accident where the cladding is exposed to high-temperature steam and can reach up to 1500°C.

The Zirconium alloy option

Stainless steels were used for the first nuclear plants but zirconium alloys (Zircaloy) have been implemented as the cladding of choice for today's reactors in order to avoid stress-corrosion cracking. More precisely, Zircaloy 4 is used for its low neutron cross-section and good corrosion performance under service conditions (around 300°C). However, Zircaloy cladding embrittlement limits the amount of energy per unit mass that can be extracted from the nuclear fuel (e.g. burnup) (1). The Nuclear Regulatory Commission has set the burnup limit to 62 MWd/kgU for that reason. Moreover, the exothermal oxidation of Zircaloy (above 1204°C) (2) is a severe limitation for beyond design basis accident conditions. If reaching this temperature, zirconium's oxidation reaction becomes autocatalytic and produces hydrogen that can potentially explode as seen in the Fukushima accident. Such limitations motivated the development of the Accident Tolerant Fuel (ATF) research program (3) with the goal of developing advanced fuel materials with improved performance under both typical and accident conditions.

The Accident Tolerant Fuel Program

By accident tolerant, we mean a fuel with larger safety margins to accidental conditions (overcoming loss of cooling in the core for a long period of time for instance) and with improved performance during normal operations (higher safety margin and/or economic efficiency). To do so, both advanced fuels and advanced claddings are considered.

With a high melting point and a high thermal conductivity (4), UN fuel would yield higher safety margins than UO₂ because for similar burn-ups the centerline fuel temperature would be lower and the melting point higher than UO₂. For the same reason, this fuel would be more accident resistant. Additionally, the higher density of uranium nitride constitutes an economic incentive since the capacity factor of the plant could be increased by having longer fuel cycles for instance.

In a similar philosophy, the melting point and thermal conductivity of UO₂ could be enhanced by adding high conductivity additive such as BeO into the fuel. This ceramic additive would conduct the heat out of the fuel rod improving the safety but also the economics.
To replace the cladding, several options are considered: develop new metallic alloys, apply coatings on the actual Zr alloy, or replace the Zr alloy with SiC. It is considered unlikely (3) that further development of the actual Zr alloy would reach the targeted 100-fold reduction in high-temperature oxidation rate. Therefore, a markedly different alloy composition should be developed but neutronic constraints largely limit the design space. A return to stainless steel cladding is date the most discussed option. Another solution consists in applying ceramic coatings such as MAX Phase or FeCrAl coatings on Zr alloys to benefit from the oxidation resistance of such materials and reduce the hydrogen pickup (5). However, this coating philosophy raises several concerns. Among others, the coating should be compliant, highly adherent all along the in-core fuel residency which implies large constraints on thermal expansion coefficients and microstructural evolution under irradiation through time (no delamination of coating spallation). A satisfactory option would be to altogether replace the Zircaloy with an accident-resistant material. Silicon carbide (SiC) was selected for several reasons.

Silicon carbide

Silicon carbide (SiC) is a promising candidate for replacing Zircaloy because of its resistance to radiation (6) (7) (8), good high-temperature thermomechanical properties (9), low neutron absorption and activation (10) and high-temperature corrosion resistance. Indeed, SiC oxidized 1000 times more slowly than Zircaloy during both nominal and accident conditions (11) (12) (13) (14). Despite these advantages, SiC’s implementation in today’s reactors is complicated by two elements: the brittle nature of the monolithic SiC (mSiC) and the difficulties to produce high quality material (purity, stoichiometry, defect content). To overcome SiC brittleness, SiC fiber-reinforced SiC matrix composites (SiC/SiC CMC) have been developed (15) based on the experience gained from investigating SiC composites for fusion energy (8) (16) and high-temperature structural components (17). Such composites allow for pseudo-ductility: instead of failing in a single catastrophic crack, the composite layer undergoes gradual matrix microcracking followed by fiber’s failure which leads to complete failure. As such, the composite macroscopically behaves as apparently plastic with a non-frangible failure mode even though the material is purely brittle. In other words, the sample fails in a single crack and experiences three orders of magnitude more strain at failure than pure monolith and, contrary to pure monolith, does not shatter into pieces. Overall, the knowledge gathered on monolith and composites layer SiC/SiC is promising, but there is, to date, limited data published (especially on experimental work) on integrated multilayer SiC/SiC prototypical designs. As such, further work is needed to apprehend the chances of success of that option.

Multilayer composites

Today, the nuclear industry tries to combine the corrosion resistance of mSiC with the mechanical performance of CMC by developing multilayer SiC cladding designs. Monolithic SiC is needed to ensure the cladding’s hermeticity, strength, and corrosion resistance (10) while the CMC is needed to allow for pseudo-ductility. Therefore two-layer designs (duplex) and three-layer designs have been envisioned (Figure 1).
First, Ben Belgacem (18) outlined a design space for SiC cladding by assessing the temperature dependence of its mechanical properties. From that, Stone (19) compared analytically performances of duplex designs and concluded that an external monolith would reduce tensile stress. Lee (20) investigated with Finite Element Analysis (FEA) both two- and three-layer designs for small modular reactor concepts.

There is overall a need for bridging the data gap between early SiC/SiC CMC development and prototypical three-layer SiC/SiC composite relevant to the nuclear industry. As such, this work experimentally assesses the performance of three-layer structures under LOCA conditions. Furthermore, simulation of this cladding performance under service and shutdown conditions will assess the readiness of this option. To that end, finite element analysis is conducted at the core’s middle height to outline the effects of temperature gradients and irradiation induced swelling and also at the fuel rod’s end where stress related issues are raised because of the endplug presence.

2. Objectives and Scope

This thesis sets two objectives: First, this work intends to assess the three-layer design performance under simulated LOCA conditions. In other words, high temperature steam oxidation, thermal shock and burst testing are performed on three different three-layer design series in order to analyze and compare thoroughly the role and performance of each design and its individual layers.

Second, this thesis assesses the mechanical performance of three-layer SiC/SiC composites in the core by FEA on two models: at core’s middle height and at the fuel endplug. This work explains and illustrates the predominance of radiative-swelling-induced stresses over mechanical and thermal loadings. It also outlines a design space for endplug implementation by illustrating the mechanical performance of three endplug designs.

To fulfill the first objective the three-layer designs’ characteristics are assessed. Each design series is referred as Plain, Tri-axial-thin and Tri-axial-thick and are further discussed in section III. Table 1 summarizes the test matrix. Each series contains four samples. The control sample will be analyzed as-received and will be mechanically loaded until failure to assess its mechanical performance. One sample will be oxidized in high temperature steam for 48h at 1400°C at a steam flow rate of 6g/min. Lastly, two samples will be quenched from 1200°C to 100°C and 90°C water. Those samples are then mechanically loaded until failure and mechanical performance is compared against the control sample. At each step a regimen of analyses and measurements are done: optical analysis and Secondary Electron Microscopy.
(SEM) analysis, Electron Dispersive Spectroscopy, weigh measurement and X-ray analysis. This last analysis allows observing the voids’ structure, shape and distribution and estimating the porosity.

Table 1: Test Matrix: Experimental Investigations

<table>
<thead>
<tr>
<th>Sample</th>
<th>As-received Analysis</th>
<th>Oxidation &amp; Analysis</th>
<th>Thermal shock &amp; Analysis</th>
<th>Mechanical testing</th>
</tr>
</thead>
<tbody>
<tr>
<td>Plain</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>+</td>
</tr>
<tr>
<td>Plain</td>
<td>+</td>
<td>+</td>
<td></td>
<td>+</td>
</tr>
<tr>
<td>Plain</td>
<td>+</td>
<td>+</td>
<td></td>
<td>+</td>
</tr>
<tr>
<td>Plain</td>
<td>+</td>
<td>+</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Tri-axial-thin</td>
<td>+</td>
<td>+</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Tri-axial-thin</td>
<td>+</td>
<td>+</td>
<td></td>
<td>+</td>
</tr>
<tr>
<td>Tri-axial-thin</td>
<td>+</td>
<td>+</td>
<td></td>
<td>+</td>
</tr>
<tr>
<td>Tri-axial-thick</td>
<td>+</td>
<td>+</td>
<td></td>
<td>+</td>
</tr>
<tr>
<td>Tri-axial-thick</td>
<td>+</td>
<td>+</td>
<td></td>
<td>+</td>
</tr>
</tbody>
</table>

An FEA mechanical test matrix is also implemented to assess the performance of a three-layer composite cladding designs during service in core. In order to capture the global behavior of the rod, models at the core’s middle height and at the fuel rod endplug are developed. Those models are evaluated under conservative service conditions (CSC) and shutdown (SD). Section IV thoroughly described those models and the boundary conditions used. Following the assessment of the stress distribution, a sensitivity analysis on the material properties and on the in-core residence time is conducted in order to set the range of validity and bounds of the analysis. This sensitivity study is conducted on the core’s middle height model only. The simulation test matrix is listed in Table 2. At last, section VII discusses an original method to minimize the peak tensile stress in the inner monolith layer.

Table 2: Test Matrix, Finite Element Analysis

<table>
<thead>
<tr>
<th>Core Middle Height</th>
<th>Conservative Service Conditions</th>
<th>Shutdown</th>
<th>Endplug Design 1</th>
<th>+</th>
<th>+</th>
</tr>
</thead>
<tbody>
<tr>
<td>Endplug Design 2</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Endplug Design 3</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Residence Time</th>
<th>BOL¹</th>
<th>EOL²</th>
<th>CSC</th>
</tr>
</thead>
<tbody>
<tr>
<td>Swelling</td>
<td>CMC under swell</td>
<td>Same swelling</td>
<td>CMC over swells</td>
</tr>
<tr>
<td>Thermal Conductivity</td>
<td>Scatter in measured values: $k_{\text{mono}} = 10 \text{ to } 20 \frac{W}{km}$; $k_{\text{cmc}} = 2 \text{ to } 10 \frac{W}{km}$</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Mechanical properties</td>
<td>Pure elastic material</td>
<td>Pure elastic material</td>
<td>Pseudo ductile material</td>
</tr>
</tbody>
</table>

¹ Beginning Of Life
² End Of Life
3. Description of the specimens

The SiC samples consist of three layers: an inner monolith layer, a fibrous layer, and an outer environmental barrier coating (EBC) layer. All the layers are formed of high purity β-phase SiC.

The inner layer aims at containing fission gases and retaining hermeticity of the fuel rod. It is fabricated by Chemical Vapor Deposition (CVD) process and constitutes 1/6 of the total wall thickness of the samples. This deposition process is mainly used to produce bulk solids of SiC. Briefly, silicon and carbon gaseous compounds react with hydrogen at high temperature and deposit onto a substrate. This technology is used mainly because very dense and pure SiC can be obtained.

The next layer is a SiC/SiC Ceramic Matrix Composite (CMC) that accounts for 2/3 of the total wall thickness of the samples. The fabrication of the composite layer can be divided into three steps: fibers production and winding, infiltration of the fiber-matrix coating (here pyrolytic carbon PyC) and finally deposition of the matrix. First off, fiber’s development is challenging because fibers should yield high purity (no Ni or O additives), near stoichiometry (no free Si or C) and have a small diameter (around 10 µm). The presence of impurities first accelerates oxidation and second, results in a different irradiation swelling behavior than monolith SiC that engenders stresses as will be discussed in chapter V. Two companies – Nippon Carbon Co. Ltd. (Nicalon fibers) and UBE Industries Ltd (Tyranno) – provide the market with the third generation of fibers: Hi-Nicalon type S and Tyranno SA3 (6) (21). Hi-Nicalon type S fibers have been used in this work. From that, tows consisting of around 500 mono-filament beta silicon carbide fibers are wrapped around the inner monolith layer to add tensile strength to the sample and allow for pseudo-ductility. The second challenge resides in the fiber-matrix interface fabrication. Its role is to optimize the bonding strength of the fiber/matrix interface in order to maintain integrity of the composite (therefore have a strong bonding) while allowing for fiber sliding motion (therefore having a weak bonding). For nuclear applications, this interface is commonly made of single layer pyrolytic carbon (22) because it provides an optimal bond strength between fibers and matrix (weak enough to deflect the cracks but strong enough to insure the material’s integrity), and offers adequate neutron transparency (in comparison to the alternative interfaces). The last step is the deposition of the inter-fiber monolith. It is deposited by Chemical Vapor Infiltration (CVI) which is the most common industrial technique for depositing SiC in geometrically complex pieces such as a fiber network. It is costly, slow and produces a substantial number of macro pores but preserves purity and is also dense (23) (24). Figure 3 compares CVD and CVI mSiC. The mechanical benefits of the composite are unfortunately often offset by the less dense and porous characteristics of the composite that ultimately manifest itself in a lower thermal conductivity. Today, the optimum weaving pattern is starting to be extensively studied by Finite Element Analysis (FEA) but is still at its early stages.

The outermost 1/6 of the total wall thickness is made of another monolith layer. This EBC layer, deposited by CVD, mainly provides hermeticity, preventing corrosion of the CMC layer. The samples used in this work were cut from long tubes to obtain smaller lengths (approximately 25 mm) suitable for testing. Because the samples exhibited two cross-sectional cut faces (one on each end), an additional thin (~ 60µm) SiC overcoat was deposited at the samples’ surface in an attempt to protect the CMC at the cut faces from corrosion. During actual use in nuclear cladding, the cross-sectional faces would not be exposed to the coolant owing to the presence of endplugs on both ends and this overcoat layer would not be present in service.

Three architectures are investigated here corresponding to two different weaving patterns and two different sets of dimensions. The first series 196 is arranged in a herringbone pattern, sometimes
referred as plain-weave pattern, where two sets of tows are interlaced with an angle of about 90°. This series is therefore referred as Tri-Plain (three-layer, Plain weave pattern).

The two other series share a second weaving pattern design made of three tows: one running axially and two crossing each other symmetrically (+/- 60 degree angle). However, they don’t have the same thicknesses as presented in Table 3. Those two series are referred as Tri-Axial-thin at Tri-Axial-thick. Figure 2 illustrates the different weaving patterns.

Table 3: Design Parameters for three and two layer specimen

<table>
<thead>
<tr>
<th>Service</th>
<th>OD</th>
<th>Thickness</th>
<th>Typical length</th>
<th>Helicoidal Pitch</th>
<th>Interlace Angle</th>
<th>Distance between // tows (D)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tri-Plain</td>
<td>13.6 mm</td>
<td>2.6 mm</td>
<td>20.5 mm</td>
<td>27 mm</td>
<td>90°</td>
<td>1.6 mm</td>
</tr>
<tr>
<td>Tri-Axial-thin</td>
<td>11.7 mm</td>
<td>1.8 mm</td>
<td>19.9 mm</td>
<td>33 mm</td>
<td>110°</td>
<td>2.9 mm</td>
</tr>
<tr>
<td>Tri-Axial-thick</td>
<td>12.8 mm</td>
<td>2.3 mm</td>
<td>20.3 mm</td>
<td>40 mm</td>
<td>100°</td>
<td>4.2 mm</td>
</tr>
</tbody>
</table>

Figure 2: Optical view of the different WEC Designs

The SEM work presented in Figure 1(left) illustrates the three-layer architecture. The three layers are easily distinguishable by their density and roughness. While the inner monolith (top right corner) and outer EBC (bottom left corner) appear dense, the composite layer contains legions of macro pores that are located at the tow intersections. The tows have an elliptical cross section (major axis: 1mm, minor axis: 200µm). The three tow directions can also be distinguished looking at the orientation of the fibers.
Figure 3 illustrates the morphological difference between CVD (left) and CVI (right) mSiC. The CVD appears to have a typical columnar structure that is commonly observed for CVD deposits (25). On the other hand, in regions where the CVI process does not fully fill the voids, this CVI produces needle-like structures on the exterior of the CVI layer. These structures have not been reported in literature, and appear to be the result of epitaxial crystal growth.

Used for the inner and outer layers, the CVD process yields highly stoichiometric SiC with minimal impurities in comparison to CVI (9). Because mSiC performances at high temperature and/or under corrosive environments are very sensitive to the microstructural quality of the SiC, the CVD process is preferred to the infiltration. Briefly, precursor gases (CH₃SiCl₃ and H₂) react onto a hot substrate and grow SiC on it (26). However, the inter-tow matrix material of the CMC cannot be deposited and shall be infiltrated. As such, this region is filled with a monolith SiC with higher defects and impurities. First, precursors are used to deposit carbon (precursor’s decomposition) onto the CMC and CH₃SiCl₃ is further infiltrated to react with the C to form SiC (26). Unfortunately, this method is costly, slow and does not yield pure, homogeneous and dense products. To illustrate that point, one can observe that the CVI mSiC thermal conductivity is almost 10 times lower than pure mSiC (27).

Figure 4 draws a more complete picture of the sample’s architecture: upper row image show the specimen cross section and the bottom row shows the lateral outer wall of the specimen Three increasing magnifications are used: *80, *350, *1200. The top left corner image has been discussed above. The two other upper row images show a closer view of the tow intersecting region and individual fibers. As explained above, the fibers have a 10-15 micron diameter. The bottom three images illustrate the roughness of the lateral outer surface of the sample that arises from the woven pattern of the tows.
Figure 4: As-received SEM views of Tri-Axial-thin

Top: cross-section, Bottom: lateral view

This master’s thesis is organized as followed: the experimental facilities are presented in Section II before results get introduced and discussed. The third chapter of this thesis presents the as-received analysis of the samples. Those are described and characterized through optical, SEM, X-Ray analyses and mechanical performances are evaluated through a burst test. Chapters IV to V cover the experimental investigations on oxidized and quenched samples respectively. Chapter VI is dedicated to the finite element analysis study. Conclusions are summarized in Chapter VII.
II. Experimental work

1. X-Ray CT Analysis

XCT analysis was performed on all samples using a GE CT120 microCT imaging system with a 50 μm voxel size at the MIT Animal Imaging and Preclinical Testing Core. To quantify the porosity, intensity maps were converted into Boolean images using image processing algorithms. Figure 5 shows a comparison between a typical image obtained from XCT and the processed Boolean image used for void analysis. The processed Boolean image stack was then analyzed to determine the local porosity defined as the ratio of voids (empty voxels) to the entire specimen area in a cross sectional slice. Scanning through the whole sample length yields the average porosity and its standard deviation.

![Figure 5: Original (left) XCT and processed (right) image slices of the Tri-Plain series](image)

2. Burst Test

In order to evaluate the hoop strength of the tubular cladding specimens, the specimens were internally pressurized by the radial expansion of a compressed polyurethane plug until complete failure of the sample. The polyurethane plug is machined to match the specimen’s inner volume while friction between the plug and the sample is reduced using zinc stearate. Strain gages (28) (Omega SGD-2/1000-DY13) were applied to the outermost surface of the sample and were aligned to be sensitive to the hoop strain. The strain signal was then amplified with a Wheatstone bridge circuitry and recorded with an Agilent 34980A data acquisition system (DAS). Overall, the uncertainty on the strain measurements reaches around 2% because of the accumulation of gage precision, amplification accuracy and DAS’s uncertainty. The stress-strain behavior was obtained by synchronizing the strain and load measurements that yield a 0.5% uncertainty. The displacement of the metallic insert was also recorded and used to illustrate the evolution of the internal pressurization with the insert penetration. The experimental technique was validated by the measurement of the Young’s modulus and Poisson’s ratio for various metallic samples with known properties and similar geometry to the specimens investigated in this work.
3. Oxidation Background & Facility

Monolith SiC oxidation in steam for the conditions of a reflood event can be summarized in two steps (29) (30). First, a protective silica (SiO$_2$) layer is formed (Equation 1). Due to the presence of water vapor, the oxidation is followed by the formation of volatile Si(OH)$_4$ (Equation 2).

**Oxidation:** \[ SiC_{(s)} + 3H_2O_{(g)} = SiO_2_{(s)} + CO_{(g)} + 3H_2_{(g)} \] **Equation 1**

**Volatilization:** \[ SiO_2_{(s)} + 2H_2O_{(g)} = Si(OH)_4_{(g)} \] **Equation 2**

Ultimately, oxidation of mSiC results in material recession and net mass loss. Oxide thickness recession follows a paralinear law as developed by Opila (11). Unlike Zircaloy, the oxidation of SiC is 3 orders of magnitude lower and is not autocatalytic. This constitutes a large advantage over Zircaloy.

The oxidation of composites is more complex because it involves the preferential reaction of the carbon interphase over the β-SiC. Also, those composite have a higher free surface density because of the multiplication of interfaces. At last, oxidation is very sensitive to departure from stoichiometry which is a parameter more difficult to control in CMC.

Several mechanisms have been proposed to explain composites oxidation at high temperature depending of the oxygen activity (23). At low oxygen activity, oxidation can be described with the Interphase Recession Mechanism. This mechanism is driven by the preferential oxidation of the carbon interphase between matrix and fibers that free up the surface between the fiber and matrix (31), (32) as the carbon is evacuated in a gaseous oxide phase. If the SiO$_2$ formation rate is too low to seal cracks or the carbon layer too thick, debonding and self-decomposition will happen. However, this phenomenon is not expected to be dominant in the reflood event because the environment is constituted of pure steam. Extended oxidation (high oxygen activity) is described by the oxidation embrittlement mechanism (OEM) (23). Formation of brittle SiO$_2$ on the exposed surface of the composite prohibits the fiber-matrix sliding and ultimately weakens the material. As such, fibers and matrix are simultaneously loaded and crack arresting features of the CMC are lost.

The motivation for this work stems from the absence of oxidation study on multilayer prototypical samples as will be realistically made for fuel cladding. Particular attention is given to the recession of the EBC, to determining the dominant mechanism for the composite oxidation and characterizing the influence of the architecture (weaving pattern, void density). Lastly, since the oxidation is simultaneous in the three layers, the question of the strength limiting feature will be raised: is the monolith’s weakening or the composite’s embrittlement limiting the performance of oxidized multilayer composites? To do so a unique facility has been built.

The facility used to perform high-temperature steam oxidation tests consists of a sealed pool of boiling water feeding steam inside a quartz tube (Figure 6). A succession of heaters heats the steam flow up to 1400˚C. Wrapped heat tape (750W) brings the steam to 500˚C. The second heating element is an 1800W pre-heater tube furnace. Finally, a 4500W sample furnace was built using 4 Molybdenum Disilicide (MoSi$_2$) heaters placed in a cylindrical refractory structure. Testing was performed by suspending the sample in this last furnace as the steam was passed across. Because the tubular samples were open on both ends, steam was allowed to pass across both the inner and outer surfaces of the samples.
Steam flow rate, sample oxidation time, and oxidation temperature were predetermined based on the testing criteria. Steam flow rate is driven by the 1800W immersed heater whose input power is controlled through a DC power supply, thereby allowing for accurate control of the steam flow rate. Maximum input power allows reaching flow rates of 20 g/min. and can be controlled to +/- 0.2 g/min over the 48h of the tests. A DC power supply was used in order to insure the mass flow rate stability over long periods of time (72h) because the daily variation of the wall voltage induced non acceptable flow rate variations. Live weight measurements of the water pool insure control of steam flow rate stability. The raw weight measurement over time is presented in Figure 7. Oxidation temperature is controlled by a feedback close-loop system enclosing all the heating elements. Temperatures (measured with K and B type thermocouples) feed PID controllers that drive pulse control modules. For the temperatures at stake (above 1000°C), heat transfer is mainly radiative which insures that temperatures within the quartz tube are effectively matching the temperatures measured at the furnace. Those modules control power input to the heaters so that temperatures can be controlled within 5°C. Finally, safety is insured by a safety interlock that shut downs all the controllers if any abnormal behavior is detected by the thermocouples. At last, the column was validated against Zy-4 high temperature steam correlations (Kawasaki, Urbanic-Heidrick, Baker-Just & Cathcart-Pawel (33) (34)).
For those experiments, samples were left for 48h in a steam flow rate of 6g/min at 1400˚C. Such conditions correspond to a laminar flow regime (29). Weight measurements and SEM analysis were performed systematically prior and post oxidation. Additionally, after testing, the burst strength of the samples was measured.

4. Description of the Thermal Shock Facility

The thermal shock experiments attempt to reproduce the conditions associated with the restoration of coolant flow following a LOCA accident. Prior to the restoration of coolant flow, the fuel rod temperature increases and remaining coolant may altogether evaporate, thereby exposing the fuel rods to high temperatures. Once the emergency core cooling system is activated, the onrush of water at near-atmospheric conditions results in severe thermal shock to the previously exposed fuel rods.

The facility used to perform high-temperature quenching of the samples consists of a quartz tube rising from a pool of water at atmospheric pressure to a high-temperature furnace capable of temperatures up to 1400˚C. Samples are suspended by an alumina rod in the furnace and allowed to reach the furnace temperature before being rapidly plunged into the pool via a pneumatic actuator (Figure 8). The thermal shock facility mimics the experimental conditions that were used to establish the current Zirconium-based cladding safety criteria given in 10 CFR 50.46 (35). The entire quenching process is recorded by a high-speed video camera (around 1000 frames per second) to gather information on the boiling behavior of the cladding during the quenching process.

Figure 7: Weight measurement for a 10g/min evaporation flow rate

Figure 8: Thermal shock apparatus (35)
III. As-Received Results

1. Porosity Results

Before quantitative analysis, the XCT 3D images allow one to see the void shape and preferential location and also the void network interconnectivity. For the Tri-Plain series, the voids were populous and uniformly distributed across the CMC. For the two other series, voids were axially elongated and were evenly spread out at height locations around the circumference of the sample. For the three cases, this distribution coincides with the imprint of the weaving pattern: tow intersections and inter-tow regions are the preferred locations of large voids (Figure 9). For the Tri-Axial series, voids were concentrated at the locations where the three tows crossed, and the axial tows promoted axially-longer voids. The Plain series resulted in a homogeneous distribution of numerous voids across the composite. X-ray images for Tri-Axial-Thin were taken post burst as-received samples. That is why a crack is visible on the left side of the central image of Figure 9.

![Tri-Plain](image1) ![Tri-Axial-Thin](image2) ![Tri-Axial-Thick](image3)

Figure 9: X-Ray Image of the three architectures

The preferential elongation of the voids was quantified by further processing of the Boolean image stack. This characterization of internal void size is to quantify the pathways available for the steam axial percolation within the CMC layer when looking at oxidation results. In other words, the height qualitatively characterizes the ease of percolation. This measure was realized by tallying the axial movement (z movement) of two fictitious particles that both follow a random walk in the (r, θ) plane and an upward and downward z movement respectively. Since those particles start from the same point, the addition of the two covered distances estimate the void height. After several attempts, the maximal height obtains is selected. That way, the axial height is effectively measured from the two tips of the voids.

Porosity results are summarized in Table 4. The average porosity for each sample series was obtained by averaging the porosity for all constituent samples. The standard deviation for each series was obtained by statistical analysis of the axial variation of porosity in all constituent samples. Similar averaging and statistics was performed to obtain series void height characterization.
Table 4: Porosity Results

<table>
<thead>
<tr>
<th></th>
<th>Porosity</th>
<th>Void Axial Height</th>
<th>Void Population</th>
</tr>
</thead>
<tbody>
<tr>
<td>Plain</td>
<td>8.1% ± 1.0%</td>
<td>2.6 mm ± 1.8 mm</td>
<td>Homogeneous &amp; Populous</td>
</tr>
<tr>
<td>Axial-Thin</td>
<td>2.5% ± 1.1%</td>
<td>1.2 mm ± 0.9 mm</td>
<td>Evenly Distributed &amp; Scarce</td>
</tr>
<tr>
<td>Axial-Thick</td>
<td>8.7% ± 2.0%</td>
<td>5.0 mm ± 3.8 mm</td>
<td>Evenly Distributed &amp; Scarce</td>
</tr>
</tbody>
</table>

Porosity values for the different series ranged from 2.5% to 8.7%. The thin design has a markedly lower porosity than the two other designs that is likely attributable to the higher efficiency of the CVI process in the much thinner composite layer. Voids in the Tri-Axial-Thin design were also the smallest: even their largest length (in the z direction) remained small as compared to the two other designs.

Void height in the Tri-Axial-thick design has the largest standard deviation because some voids are interconnected. As such voids height ranges from the typical height of an isolated void (1mm) up to 3-4 times a single isolated void’s height.

2. Mechanical Results

Figure 10 presents the evolution of the inner pressure vs. the metal insert’s displacement for all the samples. While the displacement is directly retrieved from the load frame, the inner pressure is calculated as: $p_{\text{load}} = \frac{F}{\Sigma}$ where $F$ is the load and $\Sigma$ the cross sectional area $\Sigma = \pi r_i^2$ where $r_i$ is the inner radius of the tubular samples.

![Figure 10: Inner pressure vs Displacement for the as-received failure test](image)

As a first approximation, the failure pressure seems to be directly proportional to the thickness of the samples as shown on Figure 11. The error bar for the thickness corresponds to the standard deviation on the measured thickness (12 measures for each point). No error bar is represented for the burst inner pressure since the 0.5% load frame accuracy gives an error of less than 0.5 MPa which is within the thickness of the dot plot.
The load monotonically increases with displacement, but a discontinuity at around 55 MPa sets the limit between two distinct loading regimes: regime I extends until the pressure discontinuity around 55 MPa and regime II extends from that point to complete failure. While Tri-Plain and Tri-Axial-Thick undergo those two regimes, series Tri-Axial-Thin failed directly at the end of the first loading regime.

At first, the specimens are intact and therefore, the first loading regime consists of the simultaneous loading of the three layers. Mechanical analysis –presented below- and optical analysis of samples loaded into regime II but before complete failure revealed that the discontinuity corresponds to the inner monolith failure. Because of its stiffness and inner positioning, the inner monolith undergoes the largest tensile hoop stresses and fails prematurely. At monolith failure, the polyurethane plug relieves its compressive stress by expanding into the crack and the metal insert load suddenly decreases. However, this load is quickly recovered as the plug is further compressed on the remainder of the monolith and composite layers. Tilting of the sample was often observed at this point because cracking lead to uneven loading and symmetry loss. In the second phase, the composite layer supported most of the load until ultimate failure of the sample.

It is likely that the Tri-Axial-Thin series did not exhibit the pseudo-ductile behavior because –having both a thin thickness and a small radius of curvature (see Table 3) - the stress level reached by the CMC upon inner monolith failure exceeded the composite failure stress level. The momentum and energy release upon monolith failure could further magnify the stresses in the composite at monolith failure if monolith and composite are well bounded. Calculations using the model developed below confirm this assumption. Additionally, series Tri-Axial-Thin had a low porosity (2% vs 8% for the two other series), resulting in more inter-tow matrix material and thus enhancing inter-tow crack propagation.

The SEM analysis summarized in Figure 12 confirms the primary interpretation of the pressure vs. displacement curve. Series Tri-Plain and Tri-Axial-Thick underwent large degradation of the monolith (the monolith is even absent for Tri-Axial-Thick because it fell into pieces) and the composite layer showed pseudo-ductility. The large strain and shearing undergone by the composite layer confirms that this layer failed after the inner monolith. Indeed, if they failed simultaneously, the crack in the CMC and in the monolith would have been continuous. Also, such a level of strain in the CMC would not be realistic for the monolith. As such, the monolith must have failed before the CMC largely strains. On the contrary, series Tri-Axial-Thin failed in a single continuous crack symptomatic of single loading regime failure.
3. Mechanical Model

The stress distribution in the three layer design can be calculated analytically with a few assumptions. First, the monolith/composite interface is assumed ideal: hoop strain and radial stress are continuous. Second, the EBC contribution to stress is negligible since it is very thin. Third, the inner monolith loses all load bearing capabilities after its initial cracking. Last, the Young’s modulus of the monolith and composite layers are assumed to be isotropic.

The model therefore consists of two thick-wall concentric cylinders with the following boundary conditions:

\[ p_i = p_{\text{load}}, \]  
\[ p_s, \] the interface pressure is to be determined (use of compatibility equation at the interface)  
\[ p_{\text{out}} = 0, \] the outer pressure is negligible in the tests.

As a reminder the generic thick-shell hoop stress equation is given in Equation 3 with \( p_i, p_o, r_i, r_o \) the inner and outer pressure, inner and outer radius respectively.

\[
\sigma_\theta(r) = \frac{p_i r_i^2 - p_o r_o^2}{r_o^2 - r_i^2} - \frac{r_o^2 r_i^2 (p_o - p_i)}{r_o^2 - r_i^2} \frac{r^2 (r_o^2 - r_i^2)}{r_o^2 - r_i^2} \]  
Equation 3

The two charging regimes observed experimentally are implemented in the following two mechanical models. First, the load is shared between the whole structure (inner monolith & composite). In the second scenario, only the composite carries the load.

**Load sharing regime**

The continuity of hoop strain at the interface allows for the calculation of the interface pressure. The thick shell equation in each layer yields Equation 4 and Equation 5 for \( r = r_s \).

\[
\sigma_{\text{mono}}(r_s) = \frac{p_i r_i^2 - p_s r_s^2}{r_o^2 - r_i^2} + \frac{r_o^2 r_i^2 (p_o - p_i)}{(r_o^2 - r_i^2)} \frac{r^2 (r_o^2 - r_i^2)}{r_o^2 - r_i^2} \]  
Equation 4

\[
\sigma_{\text{fiber}}(r_s) = \frac{p_s (r_o^2 + r_i^2)}{r_o^2 - r_i^2} \]  
Equation 5

Hooke’s law coupled with continuity of hoop strain yields Equation 6 & Equation 7 at the interface \( (r = r_s) \).
\[ \varepsilon_{mono} = \varepsilon_{fiber} \]  
Equation 6

\[ \frac{1}{E_{fiber}} \frac{P_s(r_o^2 + r_s^2)}{r_o^2 - r_s^2} = \frac{1}{E_{mono}} \left( \frac{p_i r_i^2 - P_s r_s^2}{r_s^2 - r_i^2} - \frac{r_i^2 (P_s - p_i)}{(r_s^2 - r_i^2)} \right) \]  
Equation 7

From where \( p_s \) can be determined in Equation 8.

\[ p_s = \frac{2r_i^2}{r_s^2 + r_i^2 + \frac{E_{mono}}{E_{fiber}} (r_s^2 - r_i^2) \left( \frac{r_s^2 + r_o^2}{r_o^2 - r_s^2} \right) p_i} \]  
Equation 8

Once \( p_s \) is known, the hoop stress at any generic position can be expressed as Equation 9 & Equation 10.

\[ \sigma_{mono} = \frac{P_s r_s^2}{r_s^2 - r_i^2} - \frac{r_i^2 r_s^2 (p_s - p_i)}{r^2 (r_o^2 - r_i^2)} \text{ for } r < r_s \]  
Equation 9

\[ \sigma_{fiber} = \frac{p_s r_s^2}{r_s^2 - r_o^2} + \frac{r_o^2 r_s^2 P_s}{r^2 (r_o^2 - r_s^2)} \text{ for } r \geq r_s \]  
Equation 10

Hoop strain is then back-calculated using Hooke’s law in Equation 11.

\[ \varepsilon_{mono} = \frac{\sigma_{mono}}{E_{mono}} \text{ and } \varepsilon_{fiber} = \frac{\sigma_{fiber}}{E_{fiber}} \]  
Equation 11

**Composite-only load regime:**

After monolith failure (regime II), the CMC carries the load on its own and the thick wall model is implemented to describe the CMC layer. The inner radius becomes \( r_s \) and \( p_s = p_i \). Thus, the strain is given in Equation 12.

\[ \sigma_\theta = \frac{P_s r_s^2}{r_s^2 - r_o^2} + \frac{r_o^2 r_s^2 p_i}{r^2 (r_o^2 - r_s^2)} \]  
Equation 12

**4. Discussion**

Strain was independently measured at the outer diameter (OD) of the Tri-Axial-Thick series specimen. Three successive loadings were performed in order to illustrate the three different loading stages of the three-layer sample (Figure 13).

Test 1 loaded the sample with the monolith intact and was terminated at the end of the elastic domain. Test 2 presents similar initial mechanical characteristics up to the yield point. Sample was loaded with the monolith intact up to its failure. The combined stress and strain jump shows the monolith failure and the transition to the composite-only loading stage. At last, Test 3 loaded the sample with the failed monolith until complete failure. The strain relief for constant stress observed before the complete failure of the sample during test 3 is likely a premise of the total failure: the development of a crack on the opposite side of the strain gage could have led to the rearrangement of the fibers and tows leading to local strain relieve at the gage. Contrary to metals, the Young’s modulus of the composites varies with increasing loading following the damage evolution in the monolith and fibers (15). Hence, the term pseudo-ductility should not be treated similarly to metals when modeling CMC mechanical performance.
Table 5 summarizes the mechanical performances for the three series. Failure stress is reported at its maximum, that is, the interior diameter (ID). The load pressure is also reported to illustrate the range of tolerable plenum pressure in core. Of particular interest is the loading condition under which hermeticity gets lost. When measured, the level of strain at the OD quickly reaches thousands of microstrains which would correspond to a non-realistic tensile hoop stress in the EBC. As such, we can assume that the EBC breaks first. Therefore, the inner monolith failure determines the loss of hermeticity. That is why this value is also reported in Table 5. Even though the plenum pressures largely overcome the nominal 14 MPa end of life plenum pressure in LWRs, SiC cladding will yield higher pressures due to enhanced fission gas release at higher temperatures.

Table 5: As-received strength testing

<table>
<thead>
<tr>
<th>Monolith Failure</th>
<th>Fiber Layer</th>
<th>Monolith Failure ID Hoop Stress</th>
<th>Fiber Layer Load pressure</th>
<th>Monolith Failure ID Hoop Stress</th>
<th>Fiber Layer Load pressure</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tri-Plain</td>
<td></td>
<td>249 MPa</td>
<td>64 MPa</td>
<td>267 MPa</td>
<td>107 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thin</td>
<td></td>
<td>267 MPa</td>
<td>54 MPa</td>
<td>115 - 232 MPa</td>
<td>54 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thick</td>
<td></td>
<td>243 MPa</td>
<td>60 MPa</td>
<td>242 MPa</td>
<td>82 MPa</td>
</tr>
</tbody>
</table>

Because the Tri-Axial-Thin sample composite layer failed right after the inner monolith, its failure stress can only be estimated between the bounds given by the two mechanical models as reported in Table 5.

The brittle character of mSiC can be probabilistically quantified using the Weibull model. In this model, failure probability is directly associated with the stress magnitude as shows in Equation 13 where the parameter m is the Weibull modulus, σ0 the characteristic stress, σc the stress below which no failure occurs, V is the volume of the specimen, Vo the reference volume.

\[
P_f(\sigma) = 1 - \exp \left( - \left( \frac{\sigma - \sigma_c}{\sigma_0} \right)^m \frac{V}{V_0} \right)
\]

Equation 13
Equation 14 is obtained by neglecting the stress below which no failure occurs for a normalized volume ($V = V_o$):

$$P_f(\sigma) = 1 - \exp\left(-\left(\frac{\sigma}{\sigma_o}\right)^m\right)$$

Equation 14

This probabilistic law is illustrated in Figure 14 with the parameters measured by Kim (36) for a Triplex design and presented in Table 6.

**Table 6: Weibull parameters for different SiC types**

<table>
<thead>
<tr>
<th></th>
<th>$m$</th>
<th>$\sigma_o$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Triplex Kim (36)</td>
<td>11.05</td>
<td>282.4 MPa</td>
</tr>
</tbody>
</table>

The values reported in Table 5 for the as-received strength (around 250 MPa) fall in the range of Kim’s measures. This model further allows calculating the maximum allowable stress given an acceptable cladding rod failure probability. In constant reduction during the last fifty years, a failure probability between $6 \times 10^{-5}$ and $10^{-6}$ has been calculated for the power plants currently in operation in 2006 (37). For a failure probability of $5 \times 10^{-6}$, the maximum hoop tensile stress is 96 MPa. Since not all the pins in the core experience the limiting conditions assumed here, this number is conservative under this point of view. However, this analysis only considers failure due to overloading and ignores various failure modes such as debris and grid to rod fretting. On that point, this analysis is non conservative.

Figure 14: Failure Probability for Three-layer CMC
IV. High Temperature Steam Experiment

1. Oxidation Results

Typical discoloration of oxidized samples was observed on their external surface as illustrated in Figure 15. This is a typical manifestation of the formation of thin oxide layers that diffract the light. This conclusion is reinforced by comprehensive SEM and EDS analyses. Silica buildup on the exterior surfaces of the sample and in the inner voids of the CMC layer was systematically observed.

Figure 15: Optical view of sample Tri-Axial-Thick pre and post oxidation

Figure 16 shows the formation of silica detected by EDS on a lateral exposed surface of an oxidized sample. The relatively large error on the EDS counts is due to both the short time (1 min) of operation and the fact that oxygen and silicon are both light elements. However, the stoichiometry of silica SiO$_2$ is correct. The corrosive environment led to the formation and superficial cracking of that silica layer on the EBC. However, the EBC served as an effective oxygen barrier and prevented radial oxygen penetration inward through the EBC toward the CMC. The cracks may have formed upon the rapid cooling of the sample after testing; explaining the efficacy of the EBC in preventing radial diffusion of oxygen to the CMC.

Figure 16: Silica formation on the exterior lateral surface after 48 hour oxidation

Silica built up in the inner CMC was located on the exposed walls of the macro pores at any depth or height in the sample. This suggests that the interconnectedness of the CMC voids allowed full axial penetration of the steam throughout the CMC, with the steam entering the CMC void network through the exposed cross-sectional faces. Figure 17 illustrates the silica growth on the inner exposed surfaces. The stoichiometry of silica is not respected here has the EDS scans both SiO$_2$ and SiC. Oxygen was never observed to ingress inward the denser intra-tow material.
A change in weight was also observed post-oxidation. For monolith SiC, weight change is a balance between oxidation (weight gain) and volatilization (weight loss) depending on the oxygen activity and oxidation time. All the samples were oxidized for 48h but the second series of samples was accidentally oxidized at a much lower steam flow rate (4 g/min instead of 6 g/min). Furthermore, the little number of tests cannot provide meaningful statistics on the real distribution of weight change. Results are summarized in Table 7. Tri-Plain and Tri-Axial series underwent the same oxidation and yield similar weight gain. It is consistent with the fact those series have a similar void fraction. The Axial-thin sample behaved however differently. A weight loss was observed. Therefore, silica’s volatilization on the outer walls of the samples overcame the weight gain in the inner CMC. This conclusion contradicts the idea that a lower mass flow rate (3.9g/min) should result in lower volatilization rate. Indeed, the lower steam flow rate fosters silica accumulation on the outer surfaces because extraction of the volatile products is more limited. However the low porosity of this sample comforts the conclusion of low silica build up: with a 2% void fraction, little silica can build up in the inner voids and the volatilization might still overcome the weight gain. The cross sectional surface – lateral (outer and inner) surface ratio was also calculated in an attempt to correlate the behavior of the three series to their geometrical parameters, but did not yield any result. Only does the porosity difference between the Tri-Axial-Thin and Tri-Axial-Thick & Tri-Plain can explain this weight difference. As a last hypothesis, a piece of material might have fall apart during or after oxidation and invalid the weight measurement. To conclude, even if results are inherently limited by the reduce number of tests; it seems that the interpretation as a balance between weight gain/loss is confirmed. As a comparison, Lee (10) has observed a normalized weight loss of -4.6 mg/cm2 on pure monolith SiC under similar conditions. This large difference between pure CVD monolith and multilayer SiC/SiC composites stems principally from the silica build up in composites.

<table>
<thead>
<tr>
<th>Steam flow rate</th>
<th>Normalized weight change</th>
<th>Void fraction</th>
</tr>
</thead>
<tbody>
<tr>
<td>Plain</td>
<td>6.1 g/min</td>
<td>+0.65 mg/cm²</td>
</tr>
<tr>
<td>Axial-thin</td>
<td>3.9 g/min</td>
<td>-0.12 mg/cm²</td>
</tr>
<tr>
<td>Axial</td>
<td>6.0 g/min</td>
<td>+0.73 mg/cm²</td>
</tr>
</tbody>
</table>
2. Mechanical Results

Figure 18 presents the comparison of as-received and oxidized samples inner pressure vs displacement curves. None of the oxidized samples experienced the second composite-only loading regime that was as-received. This suggests that pseudo-ductility is lost during oxidation. Overall, oxidized samples failed at markedly lower inner pressures than the control samples since their failure inner pressure was in the range of the inner monolith failure pressure. Table 10 compares this failure pressure to the monolith failure pressure obtained as-received. Because Tri-Axial-thick as-received sample was loaded several times as explained in the as-received analysis, the final loading curve (plotted here) does not exhibit the inner monolith failure. If this sample has been failed in one unique loading, the plot would likely be similar to the Tri-Plain as-received loading plot.

![Graph](image)

**Figure 18: Internal Pressure vs Metal insert displacement for as-received and oxidized samples**

<table>
<thead>
<tr>
<th></th>
<th>Monolith Failure</th>
<th>CMC Failure</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tri-Plain, as-received</td>
<td>64 MPa</td>
<td>107 MPa</td>
</tr>
<tr>
<td>Tri-Plain, oxidized</td>
<td>49 MPa</td>
<td></td>
</tr>
<tr>
<td>Tri-Axial-Thin, as-received</td>
<td>54 MPa</td>
<td></td>
</tr>
<tr>
<td>Tri-Axial-Thin, oxidized</td>
<td>39 MPa</td>
<td></td>
</tr>
<tr>
<td>Tri-Axial-Thick, as-received</td>
<td>60 MPa</td>
<td>82 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thick oxidized</td>
<td>68 MPa</td>
<td></td>
</tr>
</tbody>
</table>

Table 8: Comparison of internal pressure at failure for as-received and oxidized samples

The embrittlement is clearly visible on the SEM fracture analysis (Figure 19). Prior to oxidation, the sample fails by damage accumulation in the matrix and gradual loading of the fibers. As a result, fibers get exposed and extensively pulled out as is illustrated for series Tri-Axial-Thick. However, post oxidation, samples do not strain anymore and fibers do not get decoupled from the matrix. Even for the Tri-Axial-thin series that did not exhibit pseudo ductility prior to oxidation because of over loading, it is clear that the crack propagates preferentially in the monolith and circumvents around the tow. That is, even if the load is too high for the composite to be effective, fibers and tows somehow deflect the cracks. After oxidation however, the crack does not differentiate fibers and monolith and cuts completely through the three-layers. This results in fibers still encased in the matrix as can be seen in the bottom right picture.
3. Discussion & Summary

The stress-strains curves obtained at the OD of the Tri-Axial-Thin sample as-received and post oxidation are presented in Figure 20. The as-received samples is presented in blue, the samples oxidized for 48 hours is in red. Total loss of pseudo ductility is observed post oxidation. However, the SiC elastic region remained unchanged with oxidation, exhibiting similar yield point and Young’s modulus. Such results indicate that the composite layer is either more sensitive, or more affected by the oxidation than the monolith layer.

As explained above, the oxidation of pure composite structures was studied and explained with two models depending on the oxygen activity: the Oxidation Embrittlement Mechanism or the Interphase Recession Mechanism. However, the SEM analysis failed at detecting oxygen at the fiber-matrix interface even with material of high enough resolution. Instead, the complete loss of pseudo-ductility for the 48 hour oxidized samples can be explained by the axial penetration of steam through
the CMC void network and resultant deposition of silica on the void surface. It was shown that silica grew on all the internal CMC voids (Figure 17). It is known (10) that the formation of silica reduces the strength of mSiC by introducing surface-based stress concentrations. Therefore, by extension the deposition of silica on the surfaces of the voids would weaken the surrounding mSiC by introducing preferential crack nucleation sites. After a 48 hour oxidation, sufficiently severe surface flaws are introduced within the CMC voids to allow for the mSiC matrix to fail immediately upon crack initiation. These surface based stress concentrators on the void surfaces were not quantified here (as they are small and difficult to image), and therefore a mechanical model confirm this effect is not possible.

![Figure 21: Crack propagation of as-received (left) and oxidized samples (right)](image)

Figure 21 illustrates the different cracking behavior between the as-received and 48 hour oxidized samples. Whereas the as-received sample (left) retained crack deflection capability in the CMC region and had a high arrested-crack density, the 48 hour oxidized sample (right) failed with a single large crack and had few observable intra-tow arrested-cracks. Also noteworthy from Figure 21 is that the matrix of the as-received sample remains intact around the fibers and the matrix retained enough integrity to be polished, but the matrix of the 48 hour oxidized sample crumbled away during sample polishing, leading to the exposed fibers observed in the figure. Such behavior confirms the conclusions of a stronger matrix as-received in comparison to oxidized.

Steam oxidation of multilayer SiC composites resulted in the formation and growth of silica on the exterior surfaces of the specimens and also on the surfaces of the CMC inner voids. Because of restricted steam flow within the CMC layer, oxidation in the CMC results in the buildup of silica in the inner walls of the pores with negligible volatilization of the oxide. As a result, the composite matrix got weakened by the introduction of surface-based stress concentration and underwent brittle fracture due to silica formation. During mechanical testing, all the oxidized samples experienced CMC failure immediately upon monolith failure without the pseudo-ductility observed in the as-received samples.

However, the conditions presented in this work might not be completely relevant to the actual blowdown accident because samples would realistically not be oxidized by the cut ends. The attempt to seal those cut ends with a thin CVD overcoat difference did not appear to be successful. This work is still useful as it explored a limiting case.
V. Thermal Shock Experiments

1. Thermal Shock Progress

For each series, two specimens were tested: once heated to 1200°C, each specimen is quenched into 100°C and 90°C water respectively. All specimens survived the thermal shocks without any visible mechanical deterioration. The inner monolith and the EBC do not show any visible sign of mechanical degradation. Similarly, the fibers’ ends also remain intact. This result contrasts with a previous study on pure β-phase monolith samples where all samples shattered when quenched above 1000°C into 100°C water. Since those two batches were not provided by the same manufacturer the quality standards might therefore be different which limits the comparability knowing that stoichiometry and defect concentration greatly affects the thermomechanical properties of SiC. All thermal shock samples were burst tested after quenching to investigate degradation of mechanical properties.

All of the samples behaved similarly during the thermal shock. First, the specimen entered the quenching water carrying a layer of air which, in addition to the high temperature, resulted in immediate film boiling. The film boiling regime lasted approximately 10 seconds, after which the Leidenfrost point was achieved and a quench front progressed axially along the sample from bottom-to-top and required approximately three seconds to fully traverse the sample’s height. As the sample rewetted, heat was transferred by nucleate boiling until complete cooling and cessation of boiling. The quench ends and the sample reaches its equilibrium temperature at the cessation of boiling. The conclusions were similar for the two thermal shock conditions tested. Figure 23 illustrates this behavior for a sample of series Tri-Plain quenched from 1200°C into 90°C water.

No micro crack development was observed during post-quench SEM inspection. Figure 22 shows that even the fibers directly exposed to the quenching water during the thermal shock displayed no apparent differences to the as-received fibers. Overall, no microstructural change in the samples attributable to thermal shock could be detected by visual and SEM analysis.

It is believed that the film boiling observed during the quenching process of the samples prevented excessive thermal stresses from developing. Because the film boiling allowed the sample to maintain a high surface temperature, the thermal gradient and therefore the thermal stress within the sample was minimized. When transitioning to nucleate boiling, surface temperature of the sample was significantly reduced and thereby increasing the thermal gradient across the sample. However, since the sample was already largely cooled during the film boiling phase, the developing thermal stresses did not reach the critical level that would damage the sample.

Figure 22: SEM view of an exposed fiber pre- (left) and post- (right) quench (series Tri-Plain)
Figure 23: Heat transfer regimes during thermal shock for Tri-Plain-Thick.
2. Mechanical Results

Figure 24 presents the internal pressure versus displacement curves for the as-received (blue) and quenched (red and green for 100°C and 90°C respectively) samples of the three series. The mechanical loading of the 90°C quenched sample of series Tri-Axial-Thick failed due to sample tilting (sample shortened for ceramographic analysis) and no result is reported below.

The overall mechanical characteristics of each series can be captured after the quench. For the Tri-Plain series, inner monolith failure and complete failure have the same range. In a similar way, the Tri-Axial-Thin series fails with little strain. Results for the Tri-Axial-Thick series are more spread out for various reasons. The as-received sample (blue line) was loaded several times as explained in Section (II.6.) and only the first loading without monolith failure is reported here. It is noteworthy to see that the slope of the as-received (blue) and 100°C quenched (red) samples are similar. The stress-strain analysis conducted on series Tri-Axial-Thick and presented below gives further insight on this series’ behavior. Additionally, quenching temperature does not show significant influence on the behavior. Overall, quenching had negligible impact on the mechanical properties of the samples since all of the quenched samples behaved similarly to the as-received samples. The thermal shock mechanical results are reported in Table 9 along with the as-received results.

Table 9: Thermal shock mechanical results

<table>
<thead>
<tr>
<th>Sample</th>
<th>Test Conditions</th>
<th>Failure Pressure</th>
<th>Failure Hoop Stress (ID)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tri-Plain</td>
<td>As-received</td>
<td>64 MPa</td>
<td>107 MPa</td>
</tr>
<tr>
<td>Tri-Plain</td>
<td>Q 100°C</td>
<td>49 MPa</td>
<td>110 MPa</td>
</tr>
<tr>
<td>Tri-Plain</td>
<td>Q 90°C</td>
<td>53 MPa</td>
<td>102 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thin</td>
<td>As-received</td>
<td>54 MPa</td>
<td>267 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thin</td>
<td>Q 100°C</td>
<td>56 MPa</td>
<td>305 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thin</td>
<td>Q 90°C</td>
<td>67 MPa</td>
<td>378 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thick</td>
<td>As-received</td>
<td>60 MPa</td>
<td>82 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thick</td>
<td>Q 100°C</td>
<td>51 MPa</td>
<td>53 MPa</td>
</tr>
<tr>
<td>Tri-Axial-Thick</td>
<td>Q 90°C</td>
<td>58 MPa</td>
<td>No Data</td>
</tr>
</tbody>
</table>
This mechanical analysis is also supported by the SEM observations of cracks illustrated in Figure 25. While series Tri-Plain (top row) and Tri-Axial-Thick (bottom row) clearly show the pseudo ductility failure mode, the Tri-Axial-Thin results (middle row) pictures little straining and similar behaviors.

Figure 25: SEM comparison of as-received (left) vs quenched (right) series failure modes
3. Discussion & Summary

Figure 26 compares the stress-strain curves of as-received and quenched samples for series Tri-Axial-Thick. First off, a similar elastic modulus was observed for both the as-received (blue) and quenched (red and green) samples. Secondly, all the samples experienced pseudo ductility even though quenched samples did not strain as much as the control sample. The problematic sample quenched at 90°C experienced the detachment of large fragments of monolith after monolith failure at the same radial location as the strain gage leading to first of all, stress concentration and second, uneven loading pattern. Therefore, the strain measurements were corrupted after monolith failure and tilting prevented successful bursting of the sample.

The thermal shock had overall limited influence on sample mechanical characteristics: all samples survived with no quench-induced micro-cracks detected during the SEM analysis. As a result, samples behaved similarly to the as-received samples. Considering the whole LOCA scenario, reflood would follow the blowdown event where cladding is oxidized. Blowdown is therefore expected to be the most challenging step of the LOCA’s sequence of events.
VI. Finite Element Analysis under service and shutdown

This section focuses on the mechanical performance of the prototypical three-layer design in core at service and shutdown. This work aims at presenting the qualitative stress fields obtained under irradiation, thermal and mechanical loadings through finite element analysis at two regions of interest: the core’s middle height and the endplug regions. Those two models are presented in the first paragraph followed by the material properties chosen. Section 3 presents the results. The sensitivity analysis then sets the bounds to this analysis and leads to the conclusion.

1. Description of the Work

Due to axisymmetry, a 2D segment \((r, Z)\) was used to simulate the 3D tube of thickness 0.7 mm, internal radius of 4.05 mm and thickness ratio among the three layers of 2.5/6.5/1 (Figure 27). Treating the cladding as isotropic might be inaccurate depending on the manufacturing techniques used for the composite layer.

a/ Core Middle Height Stress Profile

The stress in the cladding located at the center of the core is simulated using a 2 cm-cladding segment subject to pressure and heat flux loadings. Because of the core’s axial symmetry, fuel and cladding at the exact half-height of the core are not expected to undergo any vertical displacement. Therefore, vertical motion of the 2 cm-cladding segment is disallowed at its basis.

![ADINA model, including loads and fixities – not to scale](figure)

Under service, this cladding section undergoes internal pressurization due to the release of fission gases, external pressurization (coolant water pressure) and heat flux from the fuel. Those three external loadings vary throughout the fuel’s residence time in core and at shutdown as explained below. The three layers have ideal interfacial contact and are all constrained on their base in the axial direction.

This model has been previously validated against analytical calculations performed in a work at MIT (10). In particular, the previous work does predict no pellets-to-clad hard contact throughout the fuel operation (10). Therefore the fuel can be satisfactorily represented by the heat flux it generates onto the inner wall of the cladding. Particular attention has been paid to the material properties. Lee (38) and ORNL (18) have shown that the thermal expansion coefficient and swelling dependence on temperature have a significant role in the stress development along the cladding thickness. Those two aspects have been incorporated into the model. This paper also pays attention to influence of pseudo ductility of the CMC layer.
b/ End plug stress profile

A second model (Figure 28) is developed to assess and minimize the stress distribution at the fuel rod ends (endplug joint region). To do so, an endplug and 20 cm of cladding material are modelled. Accounting for 20 cm of cladding permits to take into account the influence of both the fuel region (heated length of 5 cm) and the plenum region (15 cm) onto the joint domain.

![Figure 28: Modelling end effects on the fuel rod – not to scale.](image)

The endplug has a truncated cone shape and is represented in grey on Figure 28. Inner and outer bases of this cone correspond to the inner and outer cross sections of the cladding. This endplug design has been selected because Khalifa’s experimental work (39) has shown success in SiC/SiC joining methodology and viability in terms of both hermeticity and strength retention. This paper will more precisely discuss the influence of the opening angle (θ) (Table 10) on the end joint region stresses. Problematically, the sharp angle between the inner cladding end and the endplug generates a stress concentration point at which FE analysis diverges. As a result, predicted stresses in this area are not useful to be compared among the various designs. The comparison will be instead conducted on line AB as shown in Figure 28. Line AB was chosen based on the observed stress profile within the simulated geometry.

<table>
<thead>
<tr>
<th>Table 10: Endplug Design Choices.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
</tr>
<tr>
<td>Design a</td>
</tr>
<tr>
<td>Opening Angle θ (°)</td>
</tr>
<tr>
<td>Joint Surface (mm²)</td>
</tr>
<tr>
<td>Design b</td>
</tr>
<tr>
<td>Opening Angle θ (°)</td>
</tr>
<tr>
<td>Joint Surface (mm²)</td>
</tr>
<tr>
<td>Design c</td>
</tr>
<tr>
<td>Opening Angle θ (°)</td>
</tr>
<tr>
<td>Joint Surface (mm²)</td>
</tr>
</tbody>
</table>

Similarly to the center core model, all the interfaces are assumed ideal and the clad is constrained in the axial direction on its base. In addition, the endplug centerline is constrained in the radial direction to preserve the symmetry.
c/ Boundary conditions

During steady-state operation, the SiC cladding experiences a total stress arising from internal and external pressurization and temperature gradients (resulting from an imposed heat flux). These have been set as boundary conditions for the two models and summarized in Table 11. Throughout fuel’s core residence’s life, plenum pressure and heat flux will evolve. Therefore, those parameters need to be assessed.

First, a coolant pressure of 15.5 MPa was chosen (typical of PWR). Heat flux was estimated according to nominal conditions. The heat flux of a PWR, rated at 3.4 GWth with 193 assemblies of 17x17 pins (25 control rods) with each active fuel pin being 3.65 m long and an inner cladding diameter of 0.008 m, is calculated as 727 kW/m². During in-core residence of the SiC cladding, different operating conditions are experienced by each pin. Hence, three different cases have been investigated as summarized in Table 11: beginning (BOL) and end (EOL) of life and a conservative scenario. A peaking factor of 1.55 is used flux at the Beginning of Life (BOL) and falls to 0.75 at the End of Life (EOL). As listed in Table 11, it is assumed that the BOL case has an initial pressure of 5 MPa, to represent the initial fill gas pressure. The pressure increase at end of life is mainly due to the release of fission gases. The inner rod pressure is commonly designed to reach around 14 MPa at the EOL for common PWRs in order to maintain the cladding in compression (15.5 MPa coolant pressure). Here, the EOL pressure was assigned to 25 MPa due to the fact that the fuel temperature with SiC cladding would be higher than the fuel temperature with Zircaloy cladding. This is firstly because the fuel-clad gap is not expected to close during the fuel cycle, increasing the thermal resistance between the fuel and the coolant and secondly because the SiC has a low thermal conductivity when irradiated. Indeed, the irradiated SiC monolith and composite’s thermal conductivities were measured in the 10-20 W/K/m and 2-10 W/K/m ranges respectively (9) (15) (40). As a comparison, Zircaloy’s thermal conductivity range from 15 to 20 W/m/K for temperatures from 300 K to 900K (41). In all cases, for steady state operation, an outer cladding temperature of 600K was prescribed. Over time, the stress level in the cladding changes mainly because of fuel rod pressurization and power level changes.

The decay heat at shutdown has been neglected and no heat flux has been accounted for. As the fuel temperature significantly drops, the plenum pressure drops as well. Thus, no additional mechanical constraint was assumed under shutdown conditions.

Table 11: Simulation Conditions for SiC Cladding Stress Calculation.

<table>
<thead>
<tr>
<th>Service</th>
<th>Hot Pin Peaking Factor</th>
<th>Heat flux (MW/m²)</th>
<th>Plenum pressure (MPa)</th>
<th>Coolant Pressure (MPa)</th>
<th>Wall Temperature (K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Beginning of life</td>
<td>1.55</td>
<td>1.12</td>
<td>5</td>
<td>15.5</td>
<td>600</td>
</tr>
<tr>
<td>End of life</td>
<td>1</td>
<td>0.72</td>
<td>25</td>
<td>15.5</td>
<td>600</td>
</tr>
<tr>
<td>Conservative</td>
<td>1.55</td>
<td>1.12</td>
<td>25</td>
<td>15.5</td>
<td>600</td>
</tr>
<tr>
<td>Shutdown</td>
<td>0</td>
<td>0</td>
<td>15.5</td>
<td>15.5</td>
<td>600</td>
</tr>
</tbody>
</table>
2. Material Properties

This section summarizes the material properties values used in this worked (Table 12).

<table>
<thead>
<tr>
<th>Material Property</th>
<th>Monolith</th>
<th>Composite</th>
</tr>
</thead>
<tbody>
<tr>
<td>Young Modulus Pa (10) (19)</td>
<td>$4.6 \times 10^{11}$</td>
<td>$2.9 \times 10^{11}$</td>
</tr>
<tr>
<td>Poisson ratio (9) (42)</td>
<td>0.21</td>
<td>0.13</td>
</tr>
<tr>
<td>Density $kg/m^3$</td>
<td>3200</td>
<td>2730</td>
</tr>
<tr>
<td>Thermal Conductivity (9) (15) (40)</td>
<td>$9.5^3$ W/(m.K)</td>
<td>$1.5^3$ W/(m.K)</td>
</tr>
<tr>
<td>CTE (43)</td>
<td>Equation 16</td>
<td>Equation 16</td>
</tr>
<tr>
<td>Swelling</td>
<td>Equation 17</td>
<td>Equation 17</td>
</tr>
<tr>
<td>Pseudo ductility (19)</td>
<td>N/A</td>
<td>PLS = 163 MPa $\epsilon_{PLS} = 0.056 %$</td>
</tr>
<tr>
<td></td>
<td></td>
<td>UTS = 404 MPa $\epsilon_{UTS} = 0.494 %$</td>
</tr>
</tbody>
</table>

Even though SiC has a high thermal conductivity as-fabricated, thermal conductivity drops with irradiation as the damage induced greatly scatters phonons. A large range of thermal conductivity is reported in the literature (9) (15) (40) due to the variety of microstructure, processing and fiber options. Values of $k_{mono} = 9.5$ W/Km and $k_{cmc} = 1.5$ W/Km were selected as a conservative baseline while the impact of thermal conductivity uncertainty on the stress distribution is quantified in a sensitivity analysis.

SiC thermal expansion coefficient (TEC) was estimated as-received from (43) because its decrease with irradiation is minor (43). The instantaneous coefficient is:

$$\alpha_{ins} \left(\frac{10^{-6}}{K}\right) = -0.7765 + 1.4350.10^{-2}T - 1.2209.10^{-5}T^2 + 3.8289.10^{-9}T^3$$  \text{ Equation 15}

From which is deduced the mean TEC with a reference temperature $T_{ref} = 298$ K:

$$\bar{\alpha} = \frac{1}{T - T_{ref}} \int_{T_{ref}}^{T} \alpha_{inst} dT$$  \text{ Equation 16}

Volumetric swelling for the CVD-SiC can be extracted from experimental work of Katoh (43) (44) in the temperature ranges of 573-1173 K. The experimental values are fitted into Equation 17 and normalized into a linear swelling strain:

$$\epsilon = \frac{1}{3} \left(0.0725 - 1.2 \cdot 10^{-4} T + 5.4 \cdot 10^{-8} \cdot T^2\right)$$  \text{ Equation 17}

To include the swelling in the finite element code ADINA, the swelling has been combined within the thermal expansion. For a reference temperature $T_{ref} = 298$K, the total dimensional changes are:

$$\epsilon = \bar{\alpha}(T_{ref}T)\left(T - T_{ref}\right) + \frac{\epsilon_{sw}}{3}$$  \text{ Equation 18}

It is therefore legitimate to define the combined expansion coefficient:

\footnotesize
\text{A sensitivity analysis is also conducted}

43
\[ \alpha_{\text{combined}} = \frac{(T - T_{\text{ref}}) \bar{\alpha} + \frac{\varepsilon_{\text{sw}}}{3}}{T - T_{\text{ref}}} \]  

Equation 19

This combined coefficient should be used carefully: at shutdown (SD), the residual swelling is function of the radiation temperature, not the SD temperature. Therefore, this equation becomes:

\[ \alpha_{\text{combined,SD}} = \frac{(T_{\text{rad}} - T_{\text{ref}}) \bar{\alpha} + \frac{\varepsilon_{\text{sw}}}{3}}{T_{\text{SD}} - T_{\text{ref}}} \]  

Equation 20

Characteristics of the stress-strain curve are also reported in Table 12. Those parameters are used to describe the composite’s pseudo ductility. The influence of pseudo ductility is discussed under the sensitivity analysis section as compared to an elastic isotropic description of SiC.

3. Results

All the cases are run based on the service steady state conservative case defined in Table 11 with the properties outlined in Table 12. As the limiting constraints, the analysis focuses on the hoop stress for the stress profile of the cladding at core’s middle height and on the axial stress for the endplug region.

Core’s Middle Height

This section discusses the stress profile through the cladding thickness and its origin looking at the detail of each hoop stress and strain components. Because the analysis is linear elastic, the superposition principle allows decomposing the stresses in three categories: thermal, mechanical, and radiative-swelling induced stresses (swell stresses) as shown in Figure 29.

The pressurization conditions put the sample under tensile mechanical stresses. Mechanical hoop stress gradually decreases along the cladding radius with a discontinuity for the composite layer because the composite is half as stiff as the monolith. Overall, mechanical stresses have a negligible contribution to the total stress.

Thermal and swell-induced stresses are directly dependent on the temperature profile plotted in Figure 30. Thermal expansion monotonously increases with temperature. As such, the hot inner layer expands relatively more than the cold outer layer, which makes each of them be respectively under compressive and tensile stresses. Again, the magnitude of the stresses is larger in the two monolith layers that are stiffer than the composite.

The swell stresses are overall the exact opposite of thermal stresses but have a larger magnitude which makes them dominant. Their assessment is further developed in Figure 30.
Figure 29: Hoop Stress Decomposition

Figure 30 shows the temperature, the thermal strain (thermal expansion + swelling) and total strain in the hoop direction and its associated hoop stress through the cladding thickness. The thermal strain profile is overall the opposite of the temperature profile because swelling’s magnitude is larger than thermal expansion. As such, the hot inner monolith swells less than the cold outer EBC. However, being at equilibrium requires the total strain to be constant through the three layers (neglecting the second order Poisson’s effect). As such, compressive and tensile stresses are needed to compensate the thermal and total strain mismatch and lead to the global stress profile plotted in Figure 30.

A few comments are noteworthy: the stress profile is almost symmetric to the exception of the inner monolith and EBC not being the same thickness which un-even the stress global balance. Second, the thermal expansion strain gradient opposes the swelling strain gradient during operation which alleviates the stresses. However, the thermal expansion strain gradient is lost upon shutdown which makes shutdown scenarios more concerning than service analysis.
Hoop Strain and Temperature profiles

Figure 30: Hoop Strain, Hoop Stress and Temperature Profiles

To account properly for swelling during shutdown, inner monolith and composite layers needed to be decomposed into narrow sections where an average swelling value based on the steady-state temperature distribution was allocated to each section. To validate this discretization, normal operation results obtained with the continuous set of material properties were reproduced with the discretized simulation. Results plotted in Figure 31 are in good agreement.

Thermal Strain vs Radial Position

Figure 31: Thermal Strain and Hoop Stress: Continuous (Blue), Discretized (Red) Properties

During operations, the thermal expansion strain gradient opposes the swelling strain gradient which alleviates the stresses. However, the thermal expansion strain gradient is lost upon shutdown leading to higher stresses. This observation motivates the study of shutdown in Figure 32. As for the service case, the stress profile is globally symmetric but the stress amplitude is larger leading to higher stresses in the inner and outer layers. For either under-swelling or equal-swelling composite, the tensile stress reached at shutdown will be the limiting parameter.
The endplug region theoretically undergoes mechanical, thermal and swell-induced stress. However, the long plenum region (15 cm) completely thermally isolates the joint region from the fuel region. As a result, the temperature is homogeneous in the region of interest and therefore thermal stresses are negligible. For the same reason, the swelling-induced stresses are negligible unless the different layers have a different swelling behavior which is assessed in the sensitivity analysis section.

Only does the mechanical stress component shape the stress profile in the joint region. Because of the higher inner pressure than coolant pressure, the clad and endplug undergo internal pressurization that results into a momentum onto the joint line. Indeed, the loading inner pressure engenders the opposite rotational motion of the two pieces. Since cladding and endplug cannot interpenetrate or split,
the inner part of the joint undergoes large tensile stress as the two pieces try to separate (going asymptotically to infinity at the sharp angle) whereas the outer part of the joint undergoes compressive stress as the two pieces push one against the other. The choice of the opening angle can directly affect this joint stress profile, which is a function of cladding thickness and plug geometry.

Lastly, since thermal and swelling effects are negligible, stresses under service or shutdown are equal in the endplug region. The axial stress (limiting stress) profile is presented in Table 13 for the three designs studied.

Table 13: Axial Stress extrema on line AB for the different plug designs

<table>
<thead>
<tr>
<th>Opening Angle θ (°)</th>
<th>7.75</th>
<th>20</th>
<th>45</th>
</tr>
</thead>
<tbody>
<tr>
<td>Monolith</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Hoop</td>
<td>Min (MPa)</td>
<td>Max (MPa)</td>
<td>Min (MPa)</td>
</tr>
<tr>
<td></td>
<td>26</td>
<td>26</td>
<td>44</td>
</tr>
<tr>
<td>Composite</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Hoop</td>
<td>7</td>
<td>12</td>
<td>-22</td>
</tr>
<tr>
<td>EBC</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Hoop</td>
<td>6</td>
<td>7</td>
<td>-30</td>
</tr>
</tbody>
</table>

Since the mechanical loading is the only component of the stress field, the stress levels at the endplug have a much lower range than in at the core’s middle height. As the opening angle increases, the joint line lies more horizontally and the compression of the endplug onto the clad end is more and more effective at the outer radius. As such, the EBC compressive stress increases (absolute value) with the opening angle. In a symmetric pattern, the tensile stress increases on the inner radius with increasing angle. Since the compressive failure stress is less constraining than the tensile failure stress, low opening angles should be selected.

Those mechanical considerations should be coupled with manufacturing and more nuclear specific considerations. Indeed, joint efficiency or leaking rate as GA (39) underlined should be integrated into the design process.

As a side note, the indirect influence of the radiative-induced swelling in the fuel region onto the plenum region can be assessed. Because of the temperature gradient, large stress fields are observed in the fuel region. However, SiC being stiff (E= 460 GPa), the stress does not propagate into the plenum region for more than a centimeter as illustrates Figure 33. This is mainly due to the temperature propagation than effective stress propagation. For reduced-size experiments (rodlets) where a plenum region is not needed because fission gases don’t build up, a centimeter of plenum region is still necessary to thermally isolate the endplug form the fuel.
4. Sensitivity Analysis

Thermal conductivity

As briefly explained above, post irradiation thermal conductivities values are scattered (9) (15) (40):

\[ k_{\text{mono}} = 10 \text{ to } 20 \frac{W}{Km} \quad ; \quad k_{\text{cmc}} = 2 \text{ to } 10 \frac{W}{Km} \]

Because thermal conductivity shapes the temperature profile in the cladding, it strongly affects thermal and swell-induced stresses and needs accurate assessment. A conductivity study was therefore conducted to quantify the impact of that parameter. Table 14 reports the maximal hoop stress experienced through the cladding under service and shutdown conditions. This maximum is located at the inner monolith wall. The temperature gradient under service is also reported to illustrate the impact of the thermal coefficient over the temperature profile.

Table 14: Thermal conductivity sensitivity study: maximal hoop stress during service and shutdown for different TC value pairs.

<table>
<thead>
<tr>
<th>( k_{\text{mono}} = 10 \text{ W/Km} )</th>
<th>( k_{\text{cmc}} ) (W/Km):</th>
<th>2</th>
<th>4</th>
<th>8</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>Service (MPa)</td>
<td>495</td>
<td>420</td>
<td>330</td>
<td>300</td>
<td></td>
</tr>
<tr>
<td>Service ΔT</td>
<td>218</td>
<td>142</td>
<td>84</td>
<td>72</td>
<td></td>
</tr>
<tr>
<td>Shutdown (MPa)</td>
<td>790</td>
<td>680</td>
<td>545</td>
<td>300</td>
<td></td>
</tr>
</tbody>
</table>

| \( k_{\text{mono}} = 20 \text{ W/Km} \) | \( k_{\text{cmc}} \) (W/Km): | 5 | 10 | 15 | 20 |
|---|---|---|---|---|
| Service (MPa) | 350 | 250 | 210 | 195 |
| Service ΔT | 105 | 58 | 43 | 35 |
| Shutdown (MPa) | 360 | 200 | 165 | 160 |
As thermal conductivity increases, the thermal gradient decreases. It results in a more homogeneous swelling behavior and global stress reduction. It is worth noting that the mechanical component of this stress (the stress due to the inner and outer pressurization) reaches around 75 MPa at service. That is, the stress induced by swelling and thermal expansion varies between roughly 110 to 250 MPa depending on the thermal conductivity model chosen. The plenum pressure reduction with increasing TC was not accounted for in this work. Indeed, as the fuel gets colder, fission gases diffusivity in the fuel decreases and plenum pressure decreases. Therefore those calculations over-estimate the mechanical constrains. This work is therefore still conservative.

Swelling

- **Core’s Middle Height**

The impact of swelling is dominant among all the material properties studied in this work, and therefore needs a precise and careful assessment.

This section shows that a marginal difference in the swelling laws of the monolith and composite can completely reshape the stress profile. This survey is further motivated by the fact that a large scatter in the data base can been observed (44) (45). Al last, the addition of sintering agents in SiC (such as boron, yttrium or aluminum) has shown to impact the magnitude of induced-irradiation swelling (46).

In addition to the baseline swelling law implemented in Equation 17, two limiting cases are considered (Equation 20 & Equation 21). They correspond to a deviation of ±25% at 500 K from Equation 17.

Figure 34 graphs the actual data set and three swelling laws.

\[
\varepsilon_{sw}^{\text{low}} = 0.0640 - 1.2 \cdot 10^{-4} T + 5.4 \cdot 10^{-8} \cdot T^2
\]

\[
\varepsilon_{sw}^{\text{large}} = 0.0775 - 1.2 \cdot 10^{-4} T + 5.4 \cdot 10^{-8} \cdot T^2
\]

Equation 21

Equation 22

**Figure 34: Swelling Sensitivity Analysis**

Reference (Red, Equation 17) Upper (Green, Equation 22) Lower (Blue, Equation 21)
From that, three cases are implemented:

1. Swelling of the CVD (Equation 17) < Swell composite (Equation 22)
2. Swelling CVD (Equation 17) = Swell composite (Equation 17)
3. Swelling CVD (Equation 17) > Swell composite (Equation 21).

The stress profiles obtained under those three scenarios are plotted in Figure 35. Having the composite under-swell allows preserving the inner monolith in compression but exposes the composite to overloading. If the composite over-swells, the inner monolith is to cumulate tensile mechanical loading and tensile radiation-swell-induced stresses and will fail.

![Figure 35: Hoop stress distribution for the three cases.](image)

*Endplug*

The scenario with the over-swelling composite does not yield realistic stresses in the middle of the core so is not further discussed at the endplug. Only is the under-swelling scenario investigated. Broadly, with the swelling mismatch, the composite is put under tension (+300 MPa) while the inner and outer monoliths are compressed (from -400 MPa to -900 MPa).

Figure 36 illustrates the axial stress (limiting stress) for the three endplug designs along line AB for an under-swelling composite. For angles between 7.75° and 20° the joint region is too far to significantly affect the stress profile of the plenum region. An increasing angle accentuates the compressive work of the endplug onto the cladding at the outer radius and alleviates the compression on the inner monolith layer. As such, the composite axial stress reaches a higher peak stress which should be avoided.
This analysis drives to the same conclusion as the reference case discussion: smaller opening angles relieve the maximal tensile stress either in the monolith (reference case), either in the composite (under-swelling composite).

**Pseudo ductility**

The principal motivation behind developing multi-layered CMC is to alleviate SiC failure mode by making use of CMC’s pseudo ductile character. This characteristic was neglected in this analysis because, even if pseudo ductility plays an important role in deciphering the failure modes of SiC, it is not obvious how this failure mode affects the stress distribution. This section sets the bounds to that hypothesis by assessing the impact of pseudo ductility on the stress distribution.

Accounting for pseudo ductility –instead of treating the material as purely elastic- has a variable impact on the cladding stress distribution. For that study where the cladding is under service steady state conditions, the influence of pseudo ductility is minimal. However, pseudo ductility might greatly reshape the stress profile especially if the composite undergoes more challenging loading conditions under accident for instance.

Figure 37 shows for instance that –when all layers swell uniformly- the implementation of pseudo ductility does not significantly change the stress distribution. However, in the case of an under-swelling composite, pseudo ductility’s limits the stress level in the composite to around its yield point. To balance out, stresses need re-arrangement. As a result, the inner monolith tensile stress gets relieved. As such, not accounting for pseudo ductility is conservative.
Because pseudo ductility is unique to each batch of composites, further investigation is needed when composites’ development will reach maturity. This preliminary analysis only intends to show qualitative tendencies.

**Evolution in time**

This section assesses the evolution of the cladding hoop stress with the fuel’s residence time in core by simulating the cladding at beginning and end of life (boundary conditions in Table 11). Results are plotted in Figure 38.

With fission gases build up, cladding undergoes increasing internal pressurization with time. The plenum pressure increase from 5 MPa at BOL to 25 MPa at EOL directly results in the addition of a tensile stress component to the profile. Secondly, as the fuel is burned, the heat flux and thermal gradients across the cladding decrease at the end of life. This decrease in heat flux alleviates the thermal and radiative stresses through the cladding. Graphically, the stress discontinuities at the layers’ interfaces get reduced closer to EOL. In other words, hoop stress gradually increases due to inner pressurization but has a narrower range because the thermal gradient is moderated.
Figure 38: Hoop stress at BOL (blue), EOL (orange) and under the CCS (grey)

The conservative case reaches the high tensile stresses in the inner monolith at EOL but does not completely picture the high compressive stresses in the EBC at BOL. The compressive stresses do however raise less concern since the uniaxial compressive failure stress of SiC has been measured between -4GPa to -7GPa (47) (48) so the EBC is not expected to fail for that reason.

5. Summary

A three-layer SiC clad design of inner radius \( r_{\text{in}} = 4.02 \text{ mm} \), thickness \( t = 0.7 \text{ mm} \) and mSiC/CMC/mSiC ratio of \([0.25;0.65;0.1]\) was assessed by finite element analysis under service and shutdown conditions. The material properties were taken at dose-saturation and temperature-dependence was accounted for irradiation induced swelling. The monolith and composite were assumed isotropic elastic and the same swelling law was implemented for the tree layers.

The stresses in the fuel region were observed to be dominated by the swell-induced constrains that themselves stem from the temperature gradient across the clad thickness. The endplug being thermally isolated, stresses originated mainly from the mechanical constrains. While stresses in the endplug are relatively low, failure of the cladding in the fuel region is certain. The limiting stress is the tensile peak in the inner monolith.

The stress calculations were found to be sensitive to thermal conductivity and swelling. The tensile stress peak can be reduced by a factor of two using the two ends of the thermal conductivity data range (from \( k_{\text{mono}} = 10 \frac{W}{Km} \) & \( k_{\text{cmc}} = 2 \frac{W}{Km} \) to \( k_{\text{mono}} = 20 \frac{W}{Km} \) & \( k_{\text{cmc}} = 10 \frac{W}{Km} \)). Similarly, changing the swelling law completely reshapes the stress profile. An over-swelling composite reinforces the tensile peak in the inner monolith while an under-swelling composite shifts the tensile peak from the inner monolith to the CMC layer. In the endplug, either the under- or over-swelling composite drastically increases the stress level by introducing swelling-induced stresses that were not existent primarily. The stress profile sensibility to the swelling behavior opens the way to a new stress optimization methodology: instead of optimizing the stress relying on geometrical factors, the stress can be shaped by playing on the strain (that is, the swelling) profile of the multilayer composite. This method is further discussed in the next section.
VII. Tensile Stress minimization of the three-layer design

1. Introduction

Discussed above, the tensile stress at the inner monolith of the three-layer design constitutes the main obstacle for the commercial implementation of SiC/SiC as nuclear fuel cladding. This section discusses an original and efficient optimization method to relax this tensile stress. It shows how controlling the strain level (through the swelling) of the different layers can partially relieve the peak tensile stress. This method is a supplement to the usual geometrical optimization of the three-layer designs: thickness, radius or positioning of the three-layers.

To tackle this issue, all previous studies have focused on optimizing the geometrical parameters of the prototypical samples. Lee (38) unsuccessfully varied the relative thickness of the three layers before turning to a duplex design consisting of an inner CMC and outer monolith (mSiC). This way, the peak tensile stress is relocated in the inner composite while significantly reducing the stress level of the monolith. Stone (19) refined the model by adding the temperature dependence of several material properties and came to the same conclusion. In another study, Lee (20) suggested using this cladding for small modular reactors (SMR) where the heat flux and by extension the thermal stresses are much lower. Today, the duplex design constitutes an acceptable solution in terms of pure mechanical stresses. However, this design does not prevent percolation of the fission gases through the composite and thus does not constitute a convincing solution. That is why this section focuses on an alternative manner of reducing the stress level in the three-layer design.

The idea consists of inserting a thin layer of a low swelling material that will result in an overall reduction of the average hoop strain of the multilayer cladding. As such, the inner monolith SiC would not reach excessively large hoop strain and therefore would not experience excessive tensile stresses. In effect, the thin layer will be put under large tensile stresses that will balanced out in the other three layers via additional slight compressive stress.

This innovative concept is tested with predictive finite element stress calculations. The conservative case at the core’s middle height presented above is used as a baseline for comparison with the various improvements. First, the validity of this concept is tested by analyzing the effects of the addition of a thin PyC interfacial layer. After presenting the results and discussing the shortcomings of this solution, a more elaborated solution is presented using Si. Lastly, a second design optimization scheme is suggested and compared to the previous results.
2. The Thin Layer Option

Pyrolytic Carbon

The first variation form the baseline design consists of replacing the inner first 25 µm of CMC by pyrolytic carbon (PyC). As such, the layer is now composed of an inner monolith in contact with PyC followed by the CMC and the EBC. The material properties used for the PyC are reported in Table 15.

Table 15: PyC & Silicon Properties

<table>
<thead>
<tr>
<th></th>
<th>PyC</th>
<th>Silicon</th>
</tr>
</thead>
<tbody>
<tr>
<td>Young Modulus</td>
<td>40 GPa (49)</td>
<td>165 GPa</td>
</tr>
<tr>
<td>Poisson ratio</td>
<td>0.23 (49)</td>
<td>0.22</td>
</tr>
<tr>
<td>Density kg/m³</td>
<td>1900 (49)</td>
<td>2330</td>
</tr>
<tr>
<td>Thermal Conductivity</td>
<td>16 W/(m.K) (49)</td>
<td>10 W/m/K</td>
</tr>
<tr>
<td>CTE</td>
<td>$\alpha_{PyC} = 5.7 \times 10^{-6}$ (50)</td>
<td>$\alpha = 2.6 \left( \frac{10^{-6}}{K} \right)$</td>
</tr>
<tr>
<td>Swelling</td>
<td>See below</td>
<td>0</td>
</tr>
</tbody>
</table>

Pyrolitic carbon (PyC) was selected in this study for its singular swelling behavior under irradiation. The radiation-induced size changes can be described in two steps illustrated in Figure 39 (51). First, the irradiation at intermediate temperatures (350°C - 800°C) and low fluences gives rise to shrinkage. Wang (52) explains this shrinkage by the densification of the inter-crystallites material. Pyrolitic carbon being composed of highly anisotropic graphite crystallites, the material in between the crystallite has a high vacancy concentration that decreases under the first irradiation. After this first densification step (shrinkage), the PyC is subject to a “considerable” (52) crystallite reorientation and growth. When the reorientation phenomenon takes over the densification, the PyC is observed to have an anisotropic swelling law.

Because the complex swelling behavior of PyC depends on as-received density, isotropic character (measured with the Bacon Isotropic Factor (52)) and processing, a large range of swelling rates are reported in the literature from $-1.4 \times 10^{-4}$ (49) to $-2 \times 10^{-2}$ (51). This last value was selected in this study to exaggerate the effect of the PyC layer and see clearly to what extend this concept is valid.

While the thin PyC layer is pulled under tension (Figure 40), the three other SiC segments undergo a slight stress relief. Unfortunately, several shortcomings are noticeable. First, the effect on the peak tensile stress is minor: the stress is reduced from 520 MPa to 507 MPa. This can be attributed to several elements. Since the carbon layer is 10 times more compliant than the mSiC, the effect of PyC onto the SiC is much lower. This could be improved by having a stiffer and thicker interfacial material. Also, most
of the compressive stresses that balance out the tensile peak of the PyC are distributed on the CMC and not on the inner mSiC because the CMC is more compliant. Second, the stress discontinuities at the PyC/SiC interfaces are such that the PyC would very likely debond and break of tensile over-loading.

A simulation was run with a 125 µm thick PyC layer in order to increase the influence of the intermediate layer. The stress decrease is larger (down to 439 MPa) but still limited. However, this solution will not be selected because this new design is much too different from the original design.

In order to explore the design space for another material, different hypothetical PyC elastic moduli are tested to see what Young's modulus would be needed for a replacing material. For an acceptable stress level of around 100 MPa, the interfacial material would require a stiffness around E=200 GPa as is summarized in Table 16. With those requirements (high stiffness, low swelling), silicon was selected to form the interfacial layer.

Table 16: Maximum Hoop stress
(inner monolith)

<table>
<thead>
<tr>
<th></th>
<th>Maximum Hoop stress [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td>No PyC</td>
<td>520 MPa</td>
</tr>
<tr>
<td>E=40 GPa</td>
<td>507 MPa</td>
</tr>
<tr>
<td>E=100 GPa</td>
<td>358 MPa</td>
</tr>
<tr>
<td>E=200 GPa</td>
<td>95 MPa</td>
</tr>
</tbody>
</table>

Silicon

The properties of silicon used in this simulation are reported in Table 15. Because interstitials are particularly stable in Si (53), those interstitials form stable loops that drastically reduce the swelling of this pure material. For the sake of simplicity, the swelling rate of Si was estimated to 0 for that simulation.

The results obtained with Si are reported in Figure 41 and compared to the baseline and the PyC cases. The selection of Si yields more convincing results than with PyC. Indeed, the SiC peak stress can
be reduced to 466 MPa and 406 MPa using thin and thick layers of silicon respectively. Also, since the swelling mismatch between the SiC and the Si is smaller than the mismatch between SiC and PyC, the tensile load of the SiC is more acceptable than in the previous PyC case. However, the stress discontinuities at the interface are still concerning and lead to the last design iteration presented in the next section.

![Stress Profile with an Si Interface](image)

**Figure 41: Stress Profile with an Si Interface**

### 3. Optimization of the CMC Layer

The results presented above successfully show the theoretical validity of having an under-swelling layer close to the peak tensile stress. However, this idea is not totally convincing since the decrease in stress is globally unsatisfactory while the excessive local stresses in the interfacial layer are dissuasive. As such, having the whole CMC that would under swell would have larger benefits in two ways. First, the local peaking tensile stress in the interfacial layer should not be present anymore. Second, the stress relief of the inner monolith should be much larger since the CMC is thick.

Oak Ridge National Lab (44) explicitly showed that SiC monolith and SiC fibers have the same swelling behavior. Indeed, since the standards for fibers' quality have improved over time, monolith SiC and fibers yield similar purity and stoichiometry. As such, fibers and monolith should swell at equal rates. However, the presence of pores in a real composite might lead to a swell decrease since the internal pore might shrink down when internal stresses develop in the CMC. It is therefore realistic to consider controlling the swelling rate of the composite layer and adjust it as long as the swelling implemented in the simulation remains in the range of measured values. This swelling control could be done by adjusting the porosity and also the shape and size of the voids. Independently, a recent study has shown that the addition of sintering agents in SiC (such as boron, yttrium or aluminum) can also impact the magnitude of induced-irradiation swelling (46). Those two arguments justify the use of a
slightly under-swelling composite layer. To illustrate that point, the composite layer swelling was set up 10% lower than its original value. The monolith and composite fittings are compared to the raw data in Figure 42.

\[ \varepsilon_{\text{monolith}} = 0.0725 - 1.2 \cdot 10^{-4} T + 5.4 \cdot 10^{-8} T^2 \quad \text{Equation 23} \]

\[ \varepsilon_{\text{composite}} = 0.0710 - 1.2 \cdot 10^{-4} T + 5.4 \cdot 10^{-8} T^2 \quad \text{Equation 24} \]

Figure 42: Swelling Experimental Data and the Monolith & Composite Fittings

With an under-swelling composite layer, the peak tensile stress is relocated from the inner monolith to the composite hottest point. From a dynamic perspective, as the composite layer under-swells more, the stress in the inner composite become less and less tensile while stresses shift from compressive to tensile in the composite layer. For these particular swelling laws (Equation 23 & Equation 24), the stress in the inner monolith falls to 242 MPa while the composite stress reaches 365 MPa. This stress reduction is much larger than what could be obtained with the intermediate PyC or Si layer and could further be optimized. Indeed, the minimal peaking point will be obtained for a swelling such that the inner monolith and inner composite points reach exactly the same stress which should be roughly \( \frac{1}{2} (365 + 242) \approx 300 \) MPa. The compressive stress increase in the EBC layer is not of major concern since SiC compressive failure stress was estimated around ~4 GPa to ~7 GPa (47) (48).
Those results are encouraging for several reasons. First, even though much more study is needed, the swelling behavior of the composite layer can be adapted to reduce global stresses. By adjusting the porosity level and the pores’ shape, it is likely that one will be able to tune the strain profile and therefore the stress profile. For instance, one can completely relocate the peaking stress in its preferential region depending on the application. Furthermore, optimizing the composite itself is independent of both the addition of a thin Si layer and geometrical optimization. Therefore, it might be possible to further decrease the stress level of the three-layer design by simultaneously optimizing all those parameters.

4. Conclusion

This study proves that the stresses in multilayer SiC/SiC composites can be minimized by optimizing the swelling characteristics of the different layers. Briefly, it is shown that the addition of a low-swelling material at the inner monolith/composite interface results in a large tensile peak in the interfacial layer and in a complementary stress reduction in the inner monolith layer. More than quantitative results, this study open the way to a new optimization design space that complements the geometrical design space.

It was discussed that the optimal material to add should swell as little as possible while being stiff (around 200 GPa). In the baseline design, the peaking tensile stress reaches 520 MPa. A thin layer of Si decreases the stress to 466 MPa. If the design is further modified and the interfacial layer reaches 125 µm, the stress goes down to 375 MPa. Under the hypothesis that the composite swelling could be adjusted to 90% of the monolith swelling, the peak stress could be split between the inner monolith and inner composite and drop to around 300 MPa. All the results are summarized in Table 17. Since an acceptable cladding rod failure frequency requires a maximum stress of around 100 MPa (54), further efforts are necessary. However, the scenario considered here is very conservative. For instance, the lowest thermal conductivities, the highest heat flux and pressurization were implemented. Also, a design could cumulate both an under swelling monolith, a thin Si layer and an optimized geometry (thickness ratio for instance). It is therefore difficult to conclude on the viability of the SiC/SiC composite cladding option with this study only.
Table 17: Peak Stress Summary Table

<table>
<thead>
<tr>
<th>Test Case</th>
<th>Peak SiC Tensile Stress</th>
</tr>
</thead>
<tbody>
<tr>
<td>Baseline case</td>
<td>520 MPa</td>
</tr>
<tr>
<td>Thin PyC</td>
<td>507 MPa</td>
</tr>
<tr>
<td>Thick PyC</td>
<td>439 MPa</td>
</tr>
<tr>
<td>Thin Si</td>
<td>466 MPa</td>
</tr>
<tr>
<td>Thick Si</td>
<td>375 MPa</td>
</tr>
<tr>
<td>Under-Swelling Composite</td>
<td>242 MPa (monolith)</td>
</tr>
<tr>
<td></td>
<td>365 (composite)</td>
</tr>
<tr>
<td>Optimized Under-Swelling Composite</td>
<td>Around 300 MPa (both)</td>
</tr>
<tr>
<td>Acceptable Peak Tensile Stress</td>
<td></td>
</tr>
<tr>
<td>Silicon Carbide</td>
<td>Around 100 MPa</td>
</tr>
<tr>
<td>PyC</td>
<td>Around 400 MPa</td>
</tr>
<tr>
<td>Si</td>
<td>Around 300 MPa</td>
</tr>
</tbody>
</table>
VIII. Conclusions

Three three-layer SiC/SiC composites have been assessed as potential nuclear fuel cladding with enhanced loss of coolant accident capabilities: a Tri-Plain design where the composite layer’s fibers are yarne in in a two tows herringbone pattern, and two Tri-Axial series (Thick and Thin) that have an additional axial tow and are distinguished by their geometrical characteristics. Mechanical testing, ceramographic analysis and X-Ray diffraction were conducted to assess the different designs’ performances as-received. To simulate the LOCA, high-temperature steam oxidation and thermal shock experiments were also conducted. The in-core performance was assessed in a separate modeling effort. The middle core’s cladding and the endplug were simulated using FEA leading to a discussion on the stresses origins and potential stress optimization methods. The study draws several conclusions:

Experimental:
The main difference between the three series is the absence of pseudo ductility for the Tri-Axial-Thin series. With a reduced thickness and curvature, the failure load was exceeded in the composite at the time the composite failed, leading to simultaneous failure of the two layers. The presence of the axial tow did not yield significant difference in the mechanical behavior most likely because samples were tested in the hoop direction.

Steam oxidation resulted in the formation and growth of silica on the exterior surfaces of the specimens and also on the surfaces of the CMC inner voids. During mechanical testing, all 48 hour oxidized samples experienced CMC failure immediately upon monolith failure without the pseudo-ductility observed in the as-received samples. Overall, oxidation resulted in the formation of silica on the inner wall of the CMC voids leading to the weakening of the monolith matrix and brittle fracture.

Thermal shock had limited influence on sample mechanical characteristics: all samples survived quenching with no significant quench-induced reduction in strength. No micro-cracks were detected during the SEM analysis post quenching.

Modeling:
Independently, the performance of a three-layer silicon carbide cladding was assessed under service and shutdown conditions at both the core’s middle height and at the closed end of the fuel rod.

No cladding with the dimensions and properties suggested in this analytical study can survive either service or shutdown conditions based upon the inner monolith maximum stress.

The sensitivity to materials properties was also assessed. Maximum tensile stress can be divided by two when considering the two ends of the existing scattered thermal conductivity data. The variance existing in the swelling data also reshapes the stress profile through the cladding. An under-swelling of the composite turned out to be most advantageous because it puts the inner monolith under compression. However, swelling mismatch severely strains the interfaces that would likely debond.

Stresses in the endplug region depend on two factors: mechanical constrains and swelling-induced stresses. Under the hypothesis of similar swelling rates of the three layers, the stress level remains very low and inner stresses do not constitute a major limitation to the endplug implementation. However, mismatch in the swelling model results in much higher stresses that endplug design option can only slightly reduce. Unless the irradiation-induced swelling in the CVD and composite can be effectively controlled, the survivability of this cladding design is close to zero under LWR conditions.

Strain control of the multilayer cladding shows promising peak tensile stress reductions. The addition of a thin interfacial layer of an under-swelling material such as PyC or Si can reduce the monolith tensile stress by 10%. With a composite that swells 10% less than the monolith, the stress is reduced by 20%.
**Future work:**

This work leaves several questions un-answered. Oxidation experiments should be repeated on samples closed on both ends in order to determine oxidation level and mechanism in a sealed composite. This would also assess the EBC performance. Due to the likely inability of oxygen to diffuse through the sample, the oxidation could be markedly different. In order to reproduce the LOCA sequence of events, samples should be quenched after oxidation in addition to being quenched as-received. With the silica in the composite voids, the thermal shock experiment might yield different results. More quantitative analysis is also need. Size of the silica modules could be evaluated in order to gain a better understanding of this oxidation-enhanced embrittlement. For instance, an oxidation stress rising factor could be defined and calculated using these data.

The modeling work can also be further developed. One could couple the geometrical stress optimization method with the method suggested here. Additionally, the benefit of pre-stressing the different layers could help making multilayer SiC/SiC composite more reliable.

On the analytical side, the X-ray CT data could be loaded into an FE solver to evaluate the stress riser and the reduction in thermal conductivity due to the presence of the pores in the composite. Overall, a better understanding of the composite swelling law with pores would be very useful for further studies.
IX. Bibliography


