A structural model for multi-layered ceramic cylinders and its application to silicon carbide cladding of light water reactor fuel

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A B S T R A C T

A thermo-mechanical model for stress distribution and Weibull statistical fracture of a multi-layered SiC cladding for LWR fuel is developed. The model is validated by comparing its results to those of the Finite Element Analysis (FEA) code ANSYS. In steady-state operation, the temperature sensitive swelling may lead to undesirable tensile stresses which is anticipated to challenge the structural integrity of the fissile-gas retaining inner layer of CVD-SiC monolith in a triple layer design with the composite being the middle layer. The stress distribution is sensitive to potential differences in the swelling of the monolith from that of the composite layer. The sensitivity is discussed in this work. A double-layered SiC cladding that employs the inner SiCf/SiC composite layer, and the outer CVD-SiC layer has also been analyzed. This SiC cladding design significantly reduces failure probability as it appropriately allocates peak tensile stresses in the inner composite while significantly reducing tensile stress levels of the CVD-SiC monolith.

1. Introduction

Silicon Carbide (SiC) is being investigated as a potential alternative for the current zirconium based alloy cladding in light water reactors (LWRs). SiC exhibits several advantages over zirconium based alloys, including excellent high temperature and irradiation tolerance [1–7], and steam oxidation resistance [8–10]. Those features are expected to enhance accident tolerance of the LWR fuel. Even in operation at a similar neutron flux, SiC cladding would absorb fewer neutrons compared to a zirconium based alloy cladding [11,12]. Its generally comparable neutronic performance implies minimal departure from the current LWR reactor core design [11,12].

Thermo-mechanical properties of SiC under irradiation have been well studied, thanks to fusion and high temperature reactor structural applications [4–7,13–16]. Yet, relevant failure mechanisms of SiC cladding should be further examined to determine its design, and operation limitations. Those safety-driven limitations are critical unknowns – at this stage of the concept development – that will determine feasibility of the use of SiC cladding for LWRs. The discussions on SiC cladding oxidation in previous studies [8–10] illuminated that the major failure modes for the SiC cladding would fundamentally depart from the failure modes of zirconium-based cladding, whose primary failure mechanisms are heavily affected by oxidation [17]. Previous SiC structural assessments assume the SiC cladding geometry as a sole composite cylinder [18], and highlight importance of accounting for temperature dependent swelling. The purpose of this study is to lay a foundation for understanding critical structural failure modes for representative multi-layered SiC cladding (as well as the single layer), and discuss their implications for fuel design and reactor operation. Specifically, in this study we develop a model for stress distribution and statistical SiC cladding fracture for multi-layered SiC cladding that can be used to provide the probability of failure of SiC cladding under steady-state operation.

Failures of load bearing structures are either of the yield-dominant or fracture-dominant (fast fracture) types. Yield-dominant strength failures involve dislocation-mediated plasticity in the material, and occur in ductile materials, including zirconium based alloys at room temperature. Unlike dislocation-mediated plasticity, a fracture dominant failure occurs before general yielding through separation of atomic planes which results in the creation of new surfaces, whereas yield-dominant failure leads to shape changes [19]. Generally, no appreciable plasticity, or only highly localized plasticity, is involved in fracture dominant failures. Fracture-dominant failures usually occur for brittle materials that lack the ability to accommodate defects generated during plastic deformation. Fracture-dominant failure is of interest in this study because it is the failure mode of silicon carbide, a brittle ceramic material.
2. Qualitative understanding of SiC cladding fracture modes

The structure of SiC cladding can be viewed in terms of its response/tolerance to possible fractures under excessive loading. Today, a three-layer laminated structure (that consists of a monolithic SiC layer, SiC–SiC fiber composite layer, and a monolithic environmental barrier coating (EBC) layer) is regarded as a potential candidate for SiC cladding design. Monolithic SiC is well known to undergo fast fracture without a sign of plasticity if stresses are excessive. SiC/SiC fiber composites exhibit rather complex modes of failures that show some degree of pseudo-ductility [14,20], which is sometimes called brittle-like (or quasi-ductile) failure [21]. In SiC–SiC fiber-composite materials, since the CVI (Chemical Vapor Infiltration) matrix is stiffer, the fiber strength is not accessed until significant matrix microcracking occurs [22]. While CVI matrix microcracking basically follows the behavior of monolithic CVD SiC, propagating cracks from the matrix eventually meet the fibers in the composite, experiencing a barrier to the continuity of propagation. At that moment, the strength of the fiber is accessed and the fibers undergo elastic strain. The composite is regarded as essentially “leaky” above the matrix cracking stress although catastrophic fracture does not occur [5]. This crack arresting mechanism by an interaction with fibers renders the SiC composite fracture toughness (∼30 MPa m\(^{1/2}\)) roughly 10 times the monolith (∼2.5–3.0 MPa m\(^{1/2}\) for CVD SiC) [4,22]. Having said this, the seemingly ductile behavior of the SiC composite due to the fiber strength, in principle, is engineered pseudo-ductility caused by proper juxtaposition of brittle materials. In that regard, it can be inferred that the SiC composites, while possessing decent fracture toughness and engineered graceful failure modes, have essentially no strain tolerance to cracking.

Fracture of the SiC cladding layers (SiC monolith/SiC composite/EBC) can occur either individually for one layer, or in company with neighboring layers. Each layer has a specific significance to performance of the cladding as well as the fuel rod. Fractures in each layer imply degradation of certain performance metrics of the cladding. Fig. 1 illustrates qualitative evaluations for SiC cladding performance with different fracture modes in terms of fission gas retention, prevention of H\(_2\)O attack of the carbon coating of the SiC fibers in the composite during steady-states & accidents, and strength margin for accidents. The integrity of SiC monolith has prime importance in fission gas retention and load sharing in SiC cladding. Improving cladding with the addition of pseudo-ductility with additional crack-arresting capability, the SiC composite has prime importance for load sharing and avoiding catastrophic shattering of the cladding. As discussed previously, the SiC composite needs to be protected from H\(_2\)O access to the carbon-coated fibers to avoid significant strength degradation of the composite, caused by consumption of part of the carbon coating by the H\(_2\)O–C reaction [23–25]. Therefore, without the presence of the monolithic SiC overcoat by either EBC or extended CVI overcoat, the SiC composite may undergo strength degradation even at temperatures under normal operating conditions. It is important to note that SiC composite has CVI overcoat extended for a certain thickness on the surface, which can function as an EBC with similar oxidation behavior of CVD SiC [8,10]. Therefore, even with a crack formed on the EBC layer, the composite may still be protected from H\(_2\)O access to the fibers by the presence of the CVI overcoat. Microcracks formed in the CVI matrix with cracked EBC would leave the fibers essentially unprotected from H\(_2\)O access, resulting in substantial strength degradation. While meeting steady-state performance requirements in terms of fission gas retention, and load sharing, the amount or even presence of allowable cracking in the inner monolith and fiber composite should be evaluated in the context of strength margin for accidents. It is important to note that for the innermost monolithic SiC, it may not be appropriate to think of fracture in terms of amount; being monolithic, it should be viewed in terms of allowing any propagating crack(s) or none. For the composite, the unused strength should be viewed as margin left for CVI matrix cracking to access the SiC fibers’ elastic limit. Being considerably thinner in comparison to the innermost monolith and the composite, the influence of EBC on strength safety margin was considered negligible in the qualitative evaluations shown in Fig. 1. Note that the current SiC monolith/SiC–SiC composite/SiC EBC cladding design is a result of a rather qualitative SiC cladding behavior; their relative thicknesses, or perhaps even the need for each layer will be quantitatively questioned by rigorous design studies in following discussions.

3. Tensile versus compressive stresses for SiC cladding fracture

The cylindrical fuel rod cladding experiences both tensile and compressive stresses during its life time. It is initially under compression as the operating coolant pressure in the reactor is higher than the internal helium fill gas pressure. As burnup proceeds, fuel
Fig. 1. Qualitative evaluation of SiC cladding (SiC monolith/SiC-fiber, CVI-SiC composite/EBC) performance with different fracture modes in terms of fission gas retention, H$_2$O attack of carbon coating for SiC fibers in the composite in steady-states & accidents, and strength margin for accidents (Note: The words good, poor, and bad should be regarded in qualitative comparative context).
rod’s internal pressure rises due to fission gas release. That and possible Pellet-Cladding Mechanical Interaction (PCMI)\(^1\) put the fuel rod under tensile stress. After discharge, due to absence of the primary coolant pressure, the fuel rods would be under tensile stresses while residing within surroundings at atmospheric pressure. Under accident conditions, the fuel cladding may experience additional sources for significant tensile stresses, due to both thermal shock and rapid depressurization of the core operating pressure. For brittle materials such as SiC, this is not the case; the strength of ceramics under compressive stress is several times larger than under tensile stress\(^{[26,27]}\). The difference is due to the distinction in microstructural resistance to crack growth and the nature of crack propagation under the two different loadings. Cracks in compression tend to be closed and propagate stably, twisting out of their original direction to propagate parallel to the axis of the loading direction\(^{[26,27]}\). Compressive fracture is not caused by rapid unstable propagation of one crack, but by the slow extension of many cracks to form a crushed zone. It is not the size of the largest crack that counts but that of the average crack size\(^{[26]}\). In contrast, cracks in tension tend to open up and propagate unstably perpendicular to the applied stress. In such a case, it is the largest crack that causes failure. Considering the comparatively high susceptibility to tensile fracture, this study primarily discusses fuel rods under tensile stresses in an attempt to identify and understand the most limiting failure mechanism of SiC cladding of LWR fuel rods.

4. Development of structural failure model for SiC cladding

4.1. Statistical treatment of brittle fracture for SiC cladding

Metal’s dislocation plasticity enables minimization of the adverse effect of any particular defect. Flow of plasticity leads to a general yielding of the material in a global sense, rendering strength to exhibit a narrow Gaussian distribution of yield and/or fracture strength with ~5% or less of standard deviation. Hence, in designing with metals, a deterministic approach is followed, along with a safety margin\(^{[28]}\). In ceramic materials, without appreciable plasticity, strength values exhibit marked dependence on specimen size and geometry and wide scatter, with a typical standard deviation of ~25%, due to randomly distributed flaw populations\(^{[28,29]}\). That is, fracture occurs in a local sense – locally concentrated stress around pre-existing flaws/defects that cannot be accommodated and therefore release their strain energy by breaking bonds, leading to creation of new surfaces. A widely-used methodology to resolve such local-dependent material fracture is done through statistical approach.

A way of parameterizing this statistical nature of ceramics failure is through the Weibull distribution, which is, unlike the Gaussian distribution, asymmetric with one tail extending more than the other. The Weibull distribution for survival probability of a material of volume \(V\), \(P_f(V)\) under an applied tensile stress \(\sigma_0\) is shown in Eq. (1)\(^{[26]}\)

\[
P_f(V) = \exp \left[ - \frac{1}{V_0} \int_V \left( \frac{\sigma(\mathbf{r}) - \sigma_0}{\sigma_0} \right)^m \, dV \right]
\]

where \(m\) is the Weibull modulus that determines the shape of the Weibull curve, \(\sigma_0\) is the characteristic stress. The parameter \(\sigma_0\), and \(m\) are obtained by fitting experimental data. The parameter \(\sigma_0\) is the threshold stress below which zero probability of failure is assumed. As a specimen size (volume) gets bigger, there is a larger chance for the specimen to have a critical flaw size – given the volume dependency of the failure probability. This is an important observation especially in consideration of SiC cladding for LWR applications. This is because it is practically difficult to perform a statistically meaningful number of SiC cladding failure tests with the real cladding dimensions. Hence, a volume extrapolation from available data of small specimen is essential. The weakest link theory\(^{[30]}\) enables the volume extrapolation of ceramic failure statistics by essentially assuming that each unit of volume \(V_0\) can be failed independently. Then for a sample with different volume \(V = nV_0\), the probability of the survival is simply \(P_f(V_0)\) multiplied \(n\) times by itself, and is implied in the integral term of Eq. (1).

Once survival probability \(P_f(V)\) is obtained, failure probability \(P_f(V)\) can be calculated by subtracting the survival probability from unity.

\[
P_f(V) = 1 - P_f(V)
\]

In Eq. (1), \(V\) can be applied to each layer in a SiC cladding with \(\sigma_0\) and \(V_0\) experimentally obtained for each layer in the cladding. This enables the evaluation of survival probability of each layer. Consequently, a statistical treatment of SiC cladding is reduced to the development of the following items for Eq. (1).

(A) Calculation of the stress fields \(\sigma(\mathbf{r})\) for a SiC cladding that consists of SiC monolith/SiC composite/EBC layers.

(B) Volume extrapolation for each layer in the SiC cladding from relevant experimentally obtained Weibull statistics parameters, \(\sigma_0\) and \(m\) for each layer material.

The following discussion will be dedicated to development of each item – (A) and (B) – separately.

4.2. Stress distribution in the multi-layered SiC cladding

The total stress in the SiC cladding is the sum of individual stresses that arise from (1) pressure loading, and (2) temperature gradients across the cladding thickness. Solution schemes utilize linear superposition of elastic load effects.

(A) Stress distribution under a constant loading due to internal and external pressure differences

The hoop stress \(\sigma(r)\), and radial stresses \(\sigma(r)\), at any position in the cladding at a distance \(r\) from the center of the fuel rod are given by the Lamé equations for each layer of the SiC cladding shown in Fig. 2.

\[
\sigma(r)_{h,i} = A_i + \frac{B_i}{r^2}, \quad R_i \leq r \leq R_{i+1}, \quad \text{for } i = 1, 2, 3
\]

\[
\sigma(r)_{r,i} = A_i - \frac{B_i}{r^2}, \quad R_i \leq r \leq R_{i+1}, \quad \text{for } i = 1, 2, 3
\]

![Fig. 2. Illustration of SiC cladding for stress field calculations.](image-url)

---

\(^1\) PCMI places the Zr-cladding in tension even at a relatively low burnup.
where $A_{1,2,3}$ and $B_{1,2}$ are constants to be determined and $R_{1,2,3,4}$ are radial positions for the SiC cladding layers shown in Fig. 2. Note that subscripts 1, 2, and 3 in Eqs. (3) and (4) denote the inner monolith, SiC composite, and EBC, respectively.

The overall system of equations has 9 unknowns, $A_1$, $A_2$, $A_3$, $B_1$, $B_2$, $B_3$, $\sigma_{1,2}$, and $\sigma_{2,3}$ presented by Eqs. (3)–(5)a–i. These unknowns can be solved by imposing nine independent boundary conditions

$$
\begin{align}
\sigma(R_1)_{r=1} &= -P_1, \\
\sigma(R_2)_{r=2} &= \sigma(R_3)_{r=3}, \\
\frac{1}{\nu_{1,2}}[\sigma(R_1)_{r=1} - \nu_{1,3}\sigma(R_1)_{r=3}] &= -\frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3} - \nu_{1,2}\sigma(R_3)_{r=2}], \\
\sigma(R_3)_{r=2} &= \sigma(R_3)_{r=3}, \\
\frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3} - \nu_{1,2}\sigma(R_3)_{r=2}] &= -\frac{1}{\nu_{1,2}}[\sigma(R_2)_{r=2} - \nu_{1,3}\sigma(R_2)_{r=3}], \\
\sigma(R_4)_{r=3} &= -P_0, \\
P_i\pi R_i^2 - P_o\pi R_o^2 &= A_1\sigma_{1,2} + A_3\sigma_{2,3}, \\
\frac{1}{\nu_{1,2}}[\sigma(R_1)_{r=1} - \nu_{1,3}\sigma(R_1)_{r=3}] &= -\frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3} - \nu_{1,2}\sigma(R_3)_{r=2}], \\
\frac{1}{\nu_{1,2}}[\sigma(R_2)_{r=2} - \nu_{1,3}\sigma(R_2)_{r=3}] &= -\frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3} - \nu_{1,2}\sigma(R_3)_{r=2}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3} - \nu_{1,2}\sigma(R_3)_{r=2}] &= -\frac{1}{\nu_{1,2}}[\sigma(R_2)_{r=2} - \nu_{1,3}\sigma(R_2)_{r=3}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_2)_{r=2} - \nu_{1,3}\sigma(R_2)_{r=3}] &= -\frac{1}{\nu_{1,2}}[\sigma(R_1)_{r=1} - \nu_{1,3}\sigma(R_1)_{r=3}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_4)_{r=3} - \nu_{1,2}\sigma(R_4)_{r=2}] &= -\frac{1}{\nu_{1,2}}[\sigma(R_4)_{r=2} - \nu_{1,3}\sigma(R_4)_{r=3}],
\end{align}

$$

\text{(5)}

\text{Eqs. (3)–(5)a–i for stress calculations under pressure loading; the radial components of thermal stresses are also continuous at the interface of any two layers with equal diametral strains, and they vanish at}

(A) Thermal stresses due to a radial temperature gradient

A method to calculate the thermal stress distributions for a laminated SiC cladding for a given radial temperature distribution $T(r)$ is presented here. An axially symmetric cylinder with a constant axial strain ($e_{zz} = e_2$), yields the following equations for thermal stresses [31]:

$$
\sigma(r)_{r=1,2,3,4} = \frac{\alpha_i E_i}{1 - \nu_i} \int \Delta T r dr + \frac{E_i}{1 - \nu_i} \left( \frac{C_i}{1 - 2\nu_i} \frac{C_i}{r^2} + \nu_i \dot{e}_0 \right),
$$

\begin{align}
R_i &\leq r < R_{i+1}, \quad i = 1, 2, 3, \quad j = I, II, III
\end{align}

$\alpha$ is the thermal expansion coefficient, $\Delta T = T(r) - T_{ref}$ is the radial temperature distribution with respect to its departure from strain-free reference temperature $T_{ref}$. $C_i$ and $C_j$ are constants to be determined with the imposed boundary conditions. Note that the elastic modulus $E$, and Poisson ratio $\nu$ are regarded isotropic.

The boundary conditions in Eq. (18) can be obtained from the radial stress by the relationship in Eq. (17). The axial stress is shown in Eq. (19) [32].

$$
\sigma(r)_{r=3,4} = \sigma(r)_{r=1,2,3} + \frac{1}{r} \frac{d\sigma(r)}{dr},
$$

$$
\sigma(r)_{r=3,4} = \frac{\alpha_i E_i}{1 - \nu_i} \int \Delta T r dr - \frac{2v_i E_i C_i}{1 - \nu_i} \left( \frac{1}{1 - 2\nu_i} \right) + \frac{2\nu_i E_i C_i}{1 - \nu_i} \left( \frac{1}{1 - 2\nu_i} \right),
$$

$\sigma_{3,4} = E_i \dot{e}_0 + \frac{2\nu_i E_i C_i}{1 - \nu_i} \left( \frac{1}{1 - 2\nu_i} \right),

\begin{align}
R_i &\leq r < R_{i+1}, \quad i = 1, 2, 3, \quad j = I, II, III
\end{align}

We have seven unknowns, $C_i, C_j, \nu_i, \nu_j, e_m, e_m, \dot{e}_0, \ddot{e}_0$. The following seven boundary conditions [20]a–g are applied to solve for the unknowns.

$$
\begin{align}
\sigma(R_1)_{r=1,2,3,4} &= 0, \\
\sigma(R_2)_{r=1,2,3,4} &= 0, \\
\frac{1}{\nu_{1,2}}[\sigma(R_1)_{r=1,2,3,4} - \nu_{1,3}\sigma(R_1)_{r=3}] &= \frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3,4} - \nu_{1,2}\sigma(R_3)_{r=2}], \\
\frac{1}{\nu_{1,2}}[\sigma(R_2)_{r=2,3,4} - \nu_{1,3}\sigma(R_2)_{r=3}] &= \frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3,4} - \nu_{1,2}\sigma(R_3)_{r=2}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_3)_{r=3,4} - \nu_{1,2}\sigma(R_3)_{r=2}] &= \frac{1}{\nu_{1,2}}[\sigma(R_2)_{r=2,3,4} - \nu_{1,3}\sigma(R_2)_{r=3}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_4)_{r=3,4} - \nu_{1,2}\sigma(R_4)_{r=2}] &= \frac{1}{\nu_{1,2}}[\sigma(R_3)_{r=3,4} - \nu_{1,3}\sigma(R_3)_{r=2}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_4)_{r=3,4} - \nu_{1,2}\sigma(R_4)_{r=2}] &= \frac{1}{\nu_{1,2}}[\sigma(R_3)_{r=3,4} - \nu_{1,3}\sigma(R_3)_{r=2}], \\
\frac{1}{\nu_{1,3}}[\sigma(R_4)_{r=3,4} - \nu_{1,2}\sigma(R_4)_{r=2}] &= \frac{1}{\nu_{1,2}}[\sigma(R_3)_{r=3,4} - \nu_{1,3}\sigma(R_3)_{r=2}],
\end{align}

$$

\text{(20)}

Note that these boundary conditions are similar to the ones in Eqs. (3)–(5)a–i for stress calculations under pressure loading; the radial components of thermal stresses are also continuous at the interface of any two layers with equal diametral strains, and they vanish at
the innermost and the outermost surfaces. Thermal expansion self-equilibrates in the axial direction by having the axial strain $\varepsilon_0$ in all three layers, which leads to zero net force as stated in Eq. (26). Applying the boundary conditions results in the following linear equation that can be solved for the unknowns in the form $A_{m}X_{m} = b_{m}$. The matrix $A$ is shown in Eq. (28), and the unknown vector $X_{m}$ and the prescribed condition vector $b_{m}$ are presented in Eq. (29).

Integrating Eq. (29) twice, the general temperature solution can be obtained as shown in Eq. (30)

$$ T(r) = C_1 \ln r + C_2 $$

where $C_1$ and $C_2$ are constants to be determined for given boundary conditions. The linear heat rate $q'$ is then obtained by taking the temperature gradient with respect to the radial direction.

$$ q' = -k(2\pi r) \frac{dT}{dr} = -2\pi C_1 k $$

Applying the boundary conditions for continuous temperature at the interfaces, and the equal linear heat rates in each layer, with a constant cladding outermost surface temperature $T(R_4) = T_0$, the temperature distribution in the three layers are found

$$ T(r)_1 = \frac{q' }{2\pi} \left[ \frac{1}{k_2} \ln \left( \frac{R_2}{r} \right) + \frac{1}{k_3} \ln \left( \frac{R_3}{R_2} \right) \right] + T_0, \quad r_1 \leq r \leq r_2, \quad R_1 \leq r_{\text{inner monolith}} \leq R_2 $$

$$ T(r)_2 = \frac{q' }{2\pi k_3} \ln \left( \frac{R_3}{R_2} \right) + T_0, \quad r_2 \leq r \leq r_{\text{composite}} \leq R_3 $$

$$ T(r)_3 = \frac{q' }{2\pi k_3} \ln \left( \frac{R_3}{r} \right) + T_0, \quad R_3 \leq r_{\text{cladding}} \leq R_4 $$

Now, we can calculate the stress distributions in the cladding with relevant SIC cladding properties summarized in Table 1.

(c) Radiation Swelling Induced Stresses

Importance of radiation swelling for SIC cladding stress was investigated by past modeling effort conducted by Ben-BelGacem et al. [18]. Radiation swelling is sensitively dependent upon temperature. Hence, with given temperature gradients, radiation swelling gradients lead to differential expansions that cause stresses. Hence, essentially temperature-dependent swelling induced

<table>
<thead>
<tr>
<th>Table 1</th>
<th>Reference SiC cladding properties for stress calculations.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Materials</td>
<td>Elastic modulus in $\theta$ direction $E_{\theta}$ (GPa)</td>
</tr>
<tr>
<td>Monolithic SiC/EBC</td>
<td>CVD-SiC</td>
</tr>
<tr>
<td></td>
<td>Sneed et al. [4]</td>
</tr>
<tr>
<td>Fiber-Reinforced Composite</td>
<td>HNLS CVI SiC/SiC</td>
</tr>
</tbody>
</table>

Irradiation-effect considered.
stresses, in principle, take the same mathematical form with thermal stresses. The Hooke's law states:

\[
\begin{align*}
\varepsilon_{i,r} &= \frac{1}{E} [\sigma_{i,r} - \nu (\sigma_{i,s} + \sigma_{s,s})] + \frac{S}{3} \\
\varepsilon_{i,s} &= \frac{1}{E} [\sigma_{i,s} - \nu (\sigma_{i,r} + \sigma_{s,r})] + \frac{S}{3} \\
\varepsilon_{s,s} &= \frac{1}{E} [\sigma_{s,s} - \nu (\sigma_{i,r} + \sigma_{i,s})] + \frac{S}{3} = \varepsilon_{0,s}
\end{align*}
\]

(35) (36) (37)

Where the subscript 's' indicates 'swelling-induced'. \( S \) is the volumetric swelling. We assume isotropic swelling, and in this case, the swelling strain in the principal direction is one-third of the volumetric swelling. Swelling exhibits saturating behavior with respect to neutron dose [13]. In Katoh's study, CVD-SiC's swelling (%) saturates within 0.1–1 dpa for temperature ranges of 200–800 °C. We fitted the temperature dependent 'dose-saturated' swelling of CVD-SiC [13], and found

\[ S_{\text{CVD-SiC}}(T) = -2.4 \times 10^{-3} T + 0.031487 \]  

(38)

where \( T \) is in Kelvin (K).

Irradiation swelling data for Hi-Nicalon\textsuperscript{TM} Type-S CVD-SiC is too scarce to enable developing its own correlation. In Katoh's study [13], a data point shows dose saturated swelling of 0.68% for 2D-HNLS/CVI SiC composite at irradiation temperature of 800 °C. Assuming identical temperature differential swelling behavior to CVD-SiC, we empirically changed the constant (the second term) of Eq. (38) to fit the experimental data point of SiC/SiC\textsubscript{f} irradiation swelling

\[ S_{\text{SiC/SiC}}(T) = -2.4 \times 10^{-3} T + 0.032556 \]  

(39)

Similarly, linear temperature-dependent swelling can be extracted from experimental data [13] for the dose dependent swelling region \( S(\text{dpa}, T) \). In such a case, exactly the same approach can be used to find temperature-swelling \( S(T) \) at the specified dose (dpa). Since swelling of SiC cladding quickly saturates shortly after beginning of irradiation, we represent the dose-saturated swelling in Eqs. (38) and (39) and assumed their applicability throughout the irradiation period.

Sensitivities of the uncertainties of SiC composite swelling behavior are parametrically investigated in Section 5.3.

Note that Eqs. (35)–(37) are essentially the same mathematical expressions for Hooke’s law of thermal stresses if \( \frac{S}{3} \) were replaced with \( \alpha \Delta T \). Hence, applying the axis-symmetry stress relation - Eq. (17), the resulting general solutions are analogous to those of thermal stresses

\[
\sigma(r,z)_{s,i} = -\frac{E_i}{(1-\nu_i)r^2} \int \frac{S(T)}{3} r dr + \frac{E_i}{1+\nu_i} \left( \frac{C_{i,i}}{1-2\nu_i} + \frac{C_{i,s}}{1-2\nu_i} + \nu_i \epsilon_0 S \right),
\]

for \( i = 1, 2, 3 \) and \( j = I, II, III \)  

(40)

\[
\sigma(r,z)_{s,j} = -\frac{E_i}{(1-\nu_j)r^2} \int \frac{S(T)}{3} r dr - \frac{E_i S(T)}{3(1-\nu_j)},
\]

\[
+ \frac{E_i}{1+\nu_j} \left( \frac{C_{i,i}}{1-2\nu_j} + \frac{C_{i,s}}{1-2\nu_j} + \nu_i \epsilon_0 S \right),
\]

for \( i = 1, 2, 3 \) and \( j = I, II, III \)  

(41)

\[
\sigma(s,i)_{s,j} = \frac{E_s}{3(1-\nu_j)} + \frac{2\nu_i E_i C_{i,i}}{(1-2\nu_j)(1+\nu_i)} + \frac{2\nu_i^2 E_s \epsilon_0 S}{(1-2\nu_j)(1+\nu_i)},
\]

for \( i = 1, 2, 3 \) and \( j = I, II, III \)  

(42)

The boundary conditions prescribed in Eqs. (20a–g) are applicable for swelling-induced stresses, with the replacement of the term \( \alpha \Delta T \) with \( \frac{S}{3} \) for the boundary conditions (20c and e). Applying the boundary conditions results in similar linear equation that can be solved for the unknowns in the form \( A_i X_i = b_i \). The matrix \( A_i \) is the same as the one for thermal stress calculation shown in Eq. (27), and the unknown vector \( X_i \), and the prescribed condition vector \( b_i \) are presented in Eq. (43).
SiC cladding. The difference in directional mechanical behavior for SiC/SiC composite is likely to be small compared to the difference in general mechanical behavior between the SiC/SiC composite and the CVD-SiC monolith.

The study by Katoh et al. [15] also investigated neutron irradiation effects on the composite tensile properties, observing reasonably narrow scatter in measured tensile modulus, whose range was 197–244 GPa, depending on neutron fluence and irradiation temperature. A series of studies observed a decreasing Young’s modulus with neutron damage in both CVD-SiC and SiC composite [4,13,15]. However, appreciable reduction in the elastic modulus occurs after a significant neutron dose, which may be not very relevant to LWR operating conditions (e.g., 460 GPa → 407 GPa, and 230 GPa → 210 GPa for the CVD-SiC, and HNLS CVI SiC/SiC composite, respectively after a few tens of DPA) [13,15]. Hence, in this study, as far as LWR application is concerned, representative elastic moduli for both CVD-SiC and HNLS CVI SiC/SiC composite SiC were adopted from non-irradiated values.

The composite thermal conductivity was found to saturate at ~1.5 W/m K in the through-thickness direction after 2.0 DPA in 450 °C [15]. Hence, for a SiC cladding that consists of 50% monolith and 50% fiber-reinforced composite in terms of thickness, a thermal conductivity of ~5.5 W/m K is expected. This estimation is in reasonably good agreement with measured thermal conductivities (~4.8 W/m K) of Triplex SiC claddings with approximately equal CVD-monolith and fiber-reinforced composite thicknesses after irradiation in the MITR under simulated LWR core conditions in terms of radiation fields, temperature, and water chemistry [3].

4.3. Validation of the model with the finite element analysis (FEM) solution, ANSYS

The laminated SiC cladding stress model results were validated by comparison with an FEM solution obtained by the commercial FEM structural analysis code, ADINA. Typical cladding thickness of 0.57 mm was used, with the relative layer thicknesses of 4.5:4.5:1 for monolith, composite, and EBC, respectively. A linear heat rate of 18 kW/m, cladding outer most surface temperature of 350 °C, a reference temperature of 22 °C, fuel rod internal pressure of 20 MPa (at end of irradiation), external pressure of 15.5 MPa, and cladding outer-most radius of 4.75 mm were used as a reference benchmark case. The mechanical axial force balance due to end caps was emulated by imposing pressure boundary conditions on the cross sectional area of the cladding cylinders, with Eq. (11) to match the mechanical force imposed by the differential pressure in the axial direction. In ANSYS, irradiation swelling was simulated using a fictitious thermal expansion coefficient that takes into account the linear combination of the thermal expansion and the swelling. The fictitious thermal expansion coefficient $\alpha'$ and the fictitious reference temperature $T'_{ref}$ can be obtained by leveraging the nature of linear elasticity as follows

$$\alpha' \left(T - T'_{ref}\right) = \alpha(T - T'_{ref}) + \frac{S}{3}$$  (45)

Where the volumetric swelling $S$ can be expressed as a linear function of temperature $T$ at a given radiation dose, $S = aT + b$ (see Eqs. (38) and (39)).

Eq. (45) can be arranged to give

$$\alpha' = \alpha + a$$  (46)

$$T'_{ref} = -\frac{aT'_{ref} + b}{-\alpha' - a}$$  (47)

Fig. 3 shows the multi-layered cladding model in ANSYS.

The result from the stress calculation method developed in this study gives a strong agreement with that of ANSYS for all principle directions as shown in Fig. 4. The demonstrated analytical capability of resolving stress fields of multi-component cylindrical cladding facilitates relying on the model for cladding structural analysis, and design. In addition, the developed model can be potentially used in/or with a fuel modeling code.
4.4. Weibull statistics for relevant SiC cladding fracture modes

Although the Weibull approach is a very popular approach to capture the statistical nature of brittle fracture, it has been subject to a number of criticisms that are rooted in the lack of clearer physical basis. Lamon stated that “The Weibull approach is pure statistics. The strength is regarded simply as a variate and handled in the same way as heuristic observations” [34]. Especially, the original Weibull statistics’ lack of addressing fracture by multi axial stresses has been the subject of criticisms; the original Weibull distribution model for brittle fracture shown in Eq. (1) is essentially a model for uniaxial stress, $\sigma(r)$. A number of studies have been conducted to upgrade the original Weibull approach to handle multi axial stress effects [34–41].

Barnett et al. [37], Almeida et al. [38], and Lamon [36], presented a simple approximation to represent multiaxial effects in statistical fracture. In those approaches, the principal stresses are assumed to act independently in the tensile stress directions. As a result, the survival probability is the product of the individual survival probability for each principal direction as follows,

$$P_f(V) = \exp \left[ -\frac{1}{V_0} \int_V \left( \frac{\sigma_i(f) - \sigma_{u,i}}{\sigma_{0,i}} + \frac{\sigma_i(f) - \sigma_{u,j}}{\sigma_{0,j}} \right)^{m_i} \, dV \right]$$

where the subscripts $i, j$, and $k$ represent different principal stress directions. In this study, Weibull parameters ($\sigma_0, m$, and $\sigma_f$) found for uniaxial loading tests were assumed to hold for the other stress directions. In this study, the Weibull threshold stress $\sigma_{0,i}$ is considered zero. With compressive radial stresses, Eq. (48) can now be rewritten for the SiC cladding applications as follows:

$$P_f(V) = \exp \left[ -\frac{1}{V_0} \int_V \left( \frac{\sigma_0(f) - \sigma_{u,i}}{\sigma_0} + \frac{\sigma_0(f) - \sigma_{u,j}}{\sigma_0} \right)^{m} \, dV \right]$$

(49)

The Weibull modulus of CVD-SiC decreases with irradiation from typical unirradiated values of 9–10 to 7–8 on average, while the irradiated CVD-SiC. However, the reduction in Weibull modulus under irradiation exhibits a wide range of scatter depending on sample sizes and test methods [4]. Newsome found $m = 9.6$ for non-irradiated CVD-SiC and 5.5–8.7 for irradiated CVD-SiC for temperature ranges from 300 to 800 °C [43]. Similar results were also reported by Sneath [4]. The controlling mechanisms for the decrease of the Weibull modulus under irradiation with neutrons and the increase in strength are not yet understood. Katoch et al. [13] found that CVD-SiC Weibull modulus obtained with miniature 3-point flexural tests are in substantially lower ranges as $m = 2.7–6.6$ with 6.6 being the unirradiated specimen. Byun’s work with miniaturized irradiated SiC specimens exhibits Weibull moduli of 5.0 and 9.7 for the same CVD-SiC material with a slight change of the Weibull modulus under irradiation with neutrons and the change in specimen dimensions [42]. Such uncertainties in Weibull modulus of CVD-SiC are transferred into cladding integrity assessment. Change of CVD-SiC Weibull modulus may leads to a significantly different consequence of probabilistic fuel failure rate assessment. Hence, extensive studies will be needed in the future to evaluate the Weibull modulus of the actual SiC cladding with respect to neutron dose, and temperature. Characteristic strength of 369 MPa, which was obtained by uniaxial internal pressurization test with tubular CVD-SiC specimens whose size is 1.02 mm in ID, 1.22 mm in OD, and 5.8 mm in height after 1.9 dpa at 1020 °C [4]. CVD-SiC SiC exhibits evidence of strengthening with irradiation [4]. Katoh et al. [13] demonstrated the insensitivity of mechanical properties of high purity SiC and its composites to neutron irradiation doses of 30–40 dpa at temperatures 300–800 °C. CVD-SiC Weibull parameters used for SiC cladding modeling in LWR are summarized in Table 2.

<table>
<thead>
<tr>
<th>Specimen type</th>
<th>Weibull modulus (m)</th>
<th>Weibull characteristic strength (MPa)</th>
<th>Effective volume ($V_e \times 10^{-10} m^3$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CVD-SiC</td>
<td>7.5</td>
<td>369</td>
<td>6.36</td>
</tr>
</tbody>
</table>

Irradiation-effect considered.

Fracture of fiber-reinforced composite occurs in two stages. First, excessive elastic strain of the CVI-matrix causes significant microcracking of the matrix, which is seen as an end of the proportional stress–strain behavior under tensile stresses. Before substantial microcracking of the matrix, the fibers’ strength is essentially not accessed. Second, after substantial matrix microcracking, the fibers become the primary load bearing structure, undergoing plastic strain, often observed as fiber-stretching until the composite as whole reaches the ultimate stresses. A propagating crack in the matrix faces strong crack-arresting mechanisms by the blocking of fibers while a mechanical crack arresting mechanism is absent for monolithic CVD-SiC. This additional crack arresting mechanism of SiC fiber composites makes cracks from a CVD matrix fracture stay local, preventing a single crack from causing a global fracture of the material. CVI matrix cracking that leads to a departure from the proportional behavior of stress–strain is essentially a dispersed fracture failure. A significant fraction of localized fractures in a CVD matrix should take place in order for a failure event to occur. In a global sense, the dispersed fracture failure mode is generally insensitive to the size of the material [44]. That is, dispersed fracture –in either metallic materials or ceramics composites – exhibits reduced volume impact on fracture. Similar analysis holds for ultimate strength of fiber-reinforced composites as well. Ultimate failure of the fiber composite should be seen as a dispersed fracture – significant accumulation of individual fractures should occur to cause a failure. Such dispersed fracture is a critically preferable feature of the fiber-reinforced composite material; it enables ceramic material to be used in large scales with size insensitivity of strength. Yet, the strength measured on small laboratory coupons is expected to be somewhat higher than that of a larger structure [45]. Indeed, such a size-dependent strength of fiber-reinforced composite ultimate failure by the involvement of dispersed fracture is reflected in its high Weibull modulus. In this study, volume dependent Weibull modeling for SiC/SiC composite is used, which might be conservative. Weibull parameters obtained with a cladding quality Hi-Nicalon™ Type-S CVD-SiC Composite (HNLS CVI) by uniaxial stress tests [15] were used as a reference in this study. Table 3 summarizes the reference Weibull parameters for relevant fracture modes of HNLS CVI composite for LWR fuel cladding.

The currently available data only allows the assessment of individual layers of the laminated SiC cladding. Hence, in this study, the cladding load failure, $P_{f, clad}$, can be found by calculating the probability that no failures in the monolith or the composite will occur in a fuel rod as follows

$$P_{f, clad} = 1 - (1 - P_{f, monolith})(1 - P_{f, ultimate\ composite})$$

(50)
where $1 - P_{\text{monolith}}$ is the survival probability of the monolith when its failure probability is $P_{\text{monolith}}$, and $1 - P_{\text{ultimate composite}}$ is the survival probability of the composite whose ultimate failure probability is $P_{\text{ultimate composite}}$. Presence of microcracking in the CVI matrix would not imply fuel failure if the inner monolith is intact and fibers still function as the load bearing structure. In this regard, failure probability in terms of reaching the ultimate strength may be of primary interest in evaluating fuel rod failure rate. The conventional PWR fuel height, 3.658 m (12 feet) was used for calculating the cladding volume.

5. Structural analysis of the multi-layered SiC cladding

5.1. Reference cases: burnup-dependent stress origins

Stress distributions in the cladding were calculated for a reference cladding design whose total thickness remains the same as the current cladding thickness of LWRs, 0.57 mm. In this study, 45%, 45% and 10% of the SiC cladding thicknesses are used as a reference fraction for the inner CVD-SiC, the SiCf/SiC composite, and the EBC thickness, respectively. SiC cladding stress characteristics are subject to fuel operating conditions. These conditions are distinctly different for low burnup fuel (at Beginning of Life (BOL)) and high burnup conditions (at End of Life (EOL)). Table 4 summarizes the reference conditions for low burnup (BOL) and high burnup (EOL) fuel.

Note that the assumed 20 MPa of internal pressurization for the stress distribution results, shown in Figs. 4–8, implies a considerably high burnup fuel rod of SiC cladding before PCMI.\(^3\)

In the beginning of life (BOL), the fuel rod plenum pressure is below the system pressure. This leads to compressive mechanical stress, which acts to reduce the stress level as shown in Fig. 5. With burnup, fission gas build-up in a fuel rod increases the stress level in the cladding. Note that the contribution of thermal stresses to the total hoop and axial stress is significant (see Fig. 5(a) and (b)). Magnitudes of thermal stresses are comparable to mechanical stresses. Although the relatively high thermal conductivity of unirradiated CVD-SiC (166 W/m K) lower the stress in that layer, and the significantly low thermal conductivity of SiCf/SiC composite (8 W/m K) leads to appreciable thermal stresses when in contact with neighboring CVD-SiC. The self-equilibrating nature of thermal stress puts the region of higher temperature in compression and the region of lower temperature in tension. Thermal stress generally increases with a steeper temperature gradient. Thus, higher conductivity and lower fuel pin power help reduce the thermal stress levels. Thermal stresses in the SiC cladding have both positive and negative effects; they act to reduce tensile stress levels in the inner monolith, while they add tensile stresses to the

<table>
<thead>
<tr>
<th>Weibull characteristic strength (MPa)</th>
<th>Weibull modulus (m)</th>
<th>Sample length (mm)</th>
<th>Sample width (mm)</th>
<th>Sample thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Proportional limit stress</td>
<td>105</td>
<td>5</td>
<td>15</td>
<td>3</td>
</tr>
<tr>
<td>Ultimate stress</td>
<td>290</td>
<td>17.5</td>
<td>15</td>
<td>3</td>
</tr>
</tbody>
</table>

*Effective volume of the composite was taken as $15 \times 3 \times 2.33$ mm$^3$, assuming a uniform stress field was applied in the tested coupon volume. Irradiation-effect considered.

Fig. 5. Hoop (a), axial (b), and radial (c) stress distributions of the reference SiC cladding design: low burnup representative conditions at (BOL): $q = 26.1$ kW/m, $T_0 = 340$ °C, $P_i = 6.2$ MPa, $P_o = 15.5$ MPa. No appreciable irradiation effect, $k$(CVD-SiC) = 166 W/m K, $k$(SiCf/SiC) = 8 W/m K.

---

\(^3\)PCMI (Pellet Cladding Mechanical Interaction) for a LWR fuel rod of SiC cladding occurs at a substantially higher burnup (>60 MWd/kgU) than occurs for zirconium, because of the almost negligible creep of SiC cladding towards the fuel pellet. This was first investigated by a study conducted by Carpenter et al. [2].
It is important to note that no appreciable swelling is anticipated during the very early period SiC clad incore residence time. In such periods, SiC cladding is anticipated to be under compressive stresses for most of its regions because of dominating compressive mechanical stresses and compressive thermal stresses in the inner monolith. Resulting tensile thermal stresses in the SiC/SiC composite is reduced thanks to compressive mechanical stresses. In BOL, the cladding sees a total radial stress, having continuous compressive stresses whose quantities are equal to internal pressure, and external pressure at the innermost and the outermost surfaces, respectively.

As SiC clad fuel rod operates to high burnup towards EOL, fuel rod internal pressure increases and fuel pin power decreases. In addition, irradiation swelling and a substantial decrease in cladding thermal conductivities take place as noted in Table 4.

When \( P_i > P_o \), the pressure loading creates tensile mechanical hoop and axial stresses in all three layers as shown in Fig. 6. Yet, the internal pressure of 20 MPa with the external pressure of 15.5 MPa does not appreciably induce stress fields in the axial direction. Although fuel pin power decreases towards EOL, thermal stresses increase because the significant drop in thermal conductivity values outweighs the preferential effect of pin power decrease. SiC cladding experiences dose-saturated swelling soon after the very beginning of its incore residence time (swelling is saturated after \( \sim 0.1 \text{ DPA} \) when the temperature range is between 200 and 400°C [13]). Temperature-dependent SiC swelling acts in the opposite direction to thermal expansion; irradiation swelling is greater at a lower temperature [13]. This results in a swelling stress field in opposite symmetry to thermal stress field. That is, irradiation swelling puts the region of higher temperature in tension and the region of lower temperature in compression as shown in Fig. 6. Note that the swelling-induced strain is significant (\( \sim \)approximately 0.3% in the principal direction), resulting in considerable stresses. As a result of the combination of swelling-induced stress, thermal stress, and mechanical stress, the inner monolith is anticipated to be under tension – the compressive induced stress, thermal stress, and mechanical stress, the inner monolith is anticipated to be under compression.

As one can note by comparing the reference low burnup case (Fig. 5) and the high burnup case (Fig. 6), stress fields of SiC cladding considerably changes with respect to incore residence time. It is important to note that dose-dependent SiC material property changes, such as thermal conductivities and swelling-induced strain quickly saturate after \( \sim 0.1 \text{ DPA} \). It implies that SiC cladding would experience the BOL conditions in Fig. 5 for only a short period of time. For the most of its incore residence time, SiC cladding would experience stress distributions similar to ones presented in Fig. 6 with appreciable swelling effects. In the middle of life (MOL), SiC cladding stress distributions are anticipated to be slightly lower than stresses shown in Fig. 6 due to lower mechanical stress level – yet mechanical stress only contributes to a small fraction of the overall stress when appreciable thermal and swelling stresses manifest. The structural integrity of SiC cladding in EOL is more limiting than that of BOL; tensile stress levels in the inner monolithic CVD-SiC may have a potential for significantly harming hermiticity of SiC cladding with high failure probabilities caused by low Weibull modulus.

5.2. Effects of relative layer thickness fractions

An increase in the inner monolith fraction leads to a progressive reduction of tensile stresses in the inner monolith at the expense of slight stress increases in the composite and EBC. Yet, as long as the composite and EBC are under compression (negative stresses), the
increase in their stresses does not directly imply a reduction in structural integrity. Large fractions of the monolithic CVD-SiC reduces temperature gradients in the SiC cladding, resulting in a reduction in stresses arising from differential temperature fields – thermal stress and swelling. With no constraints for laminated layers, and higher thermal conductivity of the monolith SiC, SiC cladding made of the sole monolithic CVD-SiC exhibits lower tensile stress levels than the multi-layer and the sole composite cladding designs. For the sole composite cladding design, thermal and swelling stresses manifest due to low composite thermal conductivity, making for a relatively quicker transition from tension to compression along the cladding radial position. By comparing multi-layer cladding stresses with the single layer cladding stresses, one can see that the single layer treatment of the multi-layer SiC cladding significantly underestimate stress levels without a physical representation of discontinuous material properties at the interfaces. One has to be careful when choosing one cladding layer design over another; apparent stress levels alone do not imply relative ranking in structural integrity. Indeed, statistical likelihood of fractures of each layer should be accounted from the view point of hermeticity, load-bearing capability, and chemical reaction resistivity. Statistical interpretation of obtained stress fields will be addressed in following discussions.

Failure probabilities for the relevant SiC cladding fracture modes for the reference cladding design with various layer fractions under the reference high burnup (EOL) conditions whose stress distributions are shown in Fig. 7 are summarized in Table 5.

As shown in Fig. 7, the most regions of SiCf/SiC composite in the triple-layered claddings stay under compression. This sets $P_{f,\text{ultimate composite}} \approx 0$ in Eq. (50). As a result, SiC cladding’s structural integrity is determined by the inner CVD-SiC fracture with $P_{f,\text{clad}} \approx P_{f,\text{monolith}}$ as can be found for the triple layered SiC designs in Table 5. The decrease of the inner CVD-SiC fracture probability with its increasing fraction is primarily because stress level in the CVD-SiC decreases. That is, the stress decrease outweighs the volume increase in determining the failure probability. While the stress level plays a key determining factor, the volume increase slightly delays the decreasing failure probability. Yet, a significant reduction of the composite thickness could also cause a departure from the pseudo-ductile behavior of the composite. Typical SiC fiber diameter size for reactor-quality fiber-reinforced SiC/SiC composite is $\sim 11 \text{ \mu m}$ [15]. We estimate that at least 10 layers of woven SiC fiber fabrics are required for pseudo-ductile behavior. A composite layer whose thickness is 10% or 20% of the typical LWR cladding thickness ($\sim 0.57$ mm) would only allow five to $\sim$10 layers of the woven SiC fiber fabrics. The crack arresting capability of the fiber reinforced SiC composite may be maximized when the woven thickness goes beyond a certain number of woven fiber layers. In addition, further reduction of the composite layer fraction could increase the vulnerability of the cladding against mechanical impact damage. Staying below certain tensile stress level, the ultimate failure probability of SiCf/SiC composite converges to ‘zero’ which can even be seen as ‘pseudo-deterministic’ behavior. The high sensitivity of the composite failure probability to stress levels is rooted in its high Weibull modulus, $m = 17.5$ (typical engineering monolithic ceramic: $m = 5$–10, ceramic composite: $m = 10$–20; metal: $m > 50$) [29]. As the Weibull modulus gives a measure of the statistical nature of material failure, a higher Weibull modulus gives increasingly higher importance to stresses closer to the characteristic stress ($\sigma_0$) in determining the failure probability, which leads to a size-insensitive strength – the primary reason that the volume change effect is not easily seen in the composite failure behavior. The EBC is anticipated to maintain its integrity under compression. The reported triple-layered cladding load failure probabilities dictated by the CVD-SiC fracture in Table 5 are considered to be beyond the acceptable ranges.

The sole monolith cladding made of CVD-SiC gives a much lower failure probability on the order of $10^{-4}$. The reason – as can be found from the stress distribution in Fig. 7 – is the lower stress levels rooted in (1) lower thermal and swelling stresses coming from higher thermal conductivity, and (2) absence of material discontinuities. Generally, however, use of a large monolithic ceramic material under tensile stress should be avoided due to its ‘volume-dependent strength’, reflected in its small Weibull modulus. That is, the fuel cladding failure rate sharply increases upon the marginal tensile stress increase. This means that the structural integrity of the cladding could be significantly challenged even with situation that involves a marginal departure from the well-controlled situation – the problem of relying on monolithic ceramic structure for tensile loading. Nevertheless, the reported lower failure probability of the sole CVD-SiC cladding illuminates structural issues and challenges of the widely considered triple-layered SiC cladding, and requires for a potential design modification.

Use of a sole composite layer is also found to give an improved cladding load failure probability compared to the triple-layered

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Table 4
Reference conditions for low burnup (BOL) and high burnup (EOL) for SiC clad fuel rod. 

<table>
<thead>
<tr>
<th>Condition</th>
<th>Low burnup (BOL)</th>
<th>High burnup (EOL)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Internal pressure (MPa)</td>
<td>6.2</td>
<td>20</td>
</tr>
<tr>
<td>Operating pressure (MPa)</td>
<td>15.5</td>
<td>15.5</td>
</tr>
<tr>
<td>Linear pin power (kW/m)</td>
<td>26.1</td>
<td>18.0</td>
</tr>
<tr>
<td>Cladding surface temperature (°C)</td>
<td>340</td>
<td>328.8</td>
</tr>
<tr>
<td>Thermal conductivity of monolithic CVD-SiC (W/m K)</td>
<td>166</td>
<td>9.5</td>
</tr>
<tr>
<td>Thermal conductivity of SiC/SiC composite (W/m K)</td>
<td>8</td>
<td>1.5</td>
</tr>
<tr>
<td>Irradiation Swelling</td>
<td>No appreciable swelling</td>
<td>Dose-saturated swelling behavior</td>
</tr>
</tbody>
</table>

* The linear pin power is extracted from reactor physics simulation for SiC clad fueled core by Broome and Kazimi [12].
showed reduced stress levels for both the hoop and axial directions due to increasing the composite thermal conductivity from 1.5 W/m K to 3.0 W/m K under the reference high burnup (EOL) condition. Doubling the conservative thermal conductivity value reduces tensile stress levels in the inner monolith at the expense of as the stress level increase in the composite and EBC shown in Fig. 8. Such a change in the stress distribution comes from a trade-off between tensile and compressive regions associated with the symmetric thermal and swelling induced stress regions. As long as the composite and EBC region stay under compression, the increase in stress levels may not imply a reduction in structural integrity. Hence, it is infeasible hermeticity of the cladding will be enhanced if an advancement of the SiC/SiC composite thermal conductivity is achieved. This highlights the importance of the composite manufacturing technology to yielding desirable composite thermal conductivity. Yet, under an actual stress loading condition, the composite thermal conductivity is expected to reach a lower ceiling value if there is significant micro-cracking of the CVI matrix. In that regard, use of the lowest thermal conductivity of the composite seems suitable for a fuel rod stress calculation.

Thus far, the SiC cladding thickness is assumed to be the same thickness as the current Zircaloy thickness, 0.57 mm. The current manufacturers of the prototype SiC cladding, however, estimate that the minimum thickness they can achieve is close to 0.8 mm. Hence, it is important to assess SiC cladding performance with the currently available manufacturing technology. This section basically follows the same presented assessment procedure and methods with the cladding thickness of 0.8 mm. The increase in cladding thickness is applied in the outward direction, maintaining the same fuel pellet diameter. A neutronics study of a PWR core with the SiC clad fuel found that the cladding thickness increase thicknesses associated with SiC/SiC properties are considered to be far greater than those of high purity CVD-SiC.

5.3. Some relevant departure from the reference cases: composite thermal conductivity uncertainties, thicker cladding, swelling uncertainties, possible pre-stress, and thermal expansion coefficient

Since a significant portion of the SiC cladding stress comes from the significant temperature gradient in the composite (for both thermal stress and temperature-dependent swelling), thermal conductivity of the composite is an important material property to the cladding structural integrity. While this study uses a conservative SiC/SiC composite conductivity for the reference value, 1.5 W/m K (Hi-Nicalon™ Type-S), relying on a study conducted by Katoh et al. [15], Katoh’s study also presents thermal conductivity around ~3.0 W/m K for a different fiber composite (SA3 type). Such a difference in the composite thermal conductivity is closely related to presence of voids in the composite, and hence strongly affected by the chemical vapor infiltration (CVI) process. Generally, increasing the voids reduces thermal conductivity of the composite. Fig. 8 shows reduced stress levels for both the hoop and axial directions.

Table 5

<table>
<thead>
<tr>
<th>Stress distributions</th>
<th>Sole monolith</th>
<th>3:6:1</th>
<th>4.5:4:5:1(Reference)</th>
<th>6:3:1</th>
<th>Sole composite</th>
</tr>
</thead>
<tbody>
<tr>
<td>Inner CVD-SiC fracture</td>
<td>0.0078</td>
<td>1</td>
<td>0.9997</td>
<td>0.4138</td>
<td>N/A</td>
</tr>
<tr>
<td>SiC/SiC ultimate failure</td>
<td>N/A</td>
<td>7.84 x 10^{-14}</td>
<td>0</td>
<td>0</td>
<td>2.61 x 10^{-4}</td>
</tr>
<tr>
<td>SiC/SiC composite matrix cracking</td>
<td>N/A</td>
<td>0.3069</td>
<td>1.6 x 10^{-4}</td>
<td>0</td>
<td>1</td>
</tr>
<tr>
<td>EBC cracking</td>
<td>N/A</td>
<td>0</td>
<td>0.9997</td>
<td>0.4138</td>
<td>N/A</td>
</tr>
<tr>
<td>Cladding load failure</td>
<td>0.0078</td>
<td>1</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Fig. 8. The effects of composite layer thermal conductivity and total cladding thickness on stress level: the reference cladding design, the reference high burnup (EOL) conditions.
of 0.23 mm (0.57 → 0.8 mm) does not affect features of the core significantly, with the small change of the hydrogen to heavy-metal ratio (H/HM) [12]. 10% of the SiC cladding thicknesses is used as a reference fraction for the EBC thickness.

Increasing the cladding thickness helps reduce mechanical stress levels induced by the pressure loading. However, at the same time, it increases thermal stresses and swelling induced stresses. For materials with a high thermal conductivity, where most of its stresses come from the mechanical loading, increasing the thickness may bring an overall positive effect. However, for materials with a low thermal conductivity, where thermal stress and swelling induced stress account for a significant contribution to the total stress, increasing its size may hurt the structural integrity of the material, as the increase in thermal and swelling stress outweighs the decrease in mechanical stresses. Fig. 8 shows that a thicker cladding (0.8 mm) leads to an increase in the tensile stress level in the inner monolith while putting the composite and EBC in a stronger compressive stresses. Structural assessment of such a change in stress distributions needs a careful evaluation. From the hermeticity point of view, the increase of tensile stress levels in the inner monolith is not desirable. Also, an increase in fission gas release that leads to a higher fuel rod internal pressurization will take place with a thicker cladding due to fuel temperature increase. However, on the other hand, stronger compressive stresses in the SiC/SiC composite and EBC may help stop propagating cracks, which may result in a better chemical resistivity and overall structural integrity. The situation could be different during accidents, for instance during a large break loss of coolant accident (LBLOCA), when fuel pin power is much lower than the steady-state (~5 % or less of the steady-state), and the mechanical loading is higher with the core depressurization. Hence, during LBLOCA stress levels in both the monolith and the composite decrease with a thicker SiC cladding. Hence it can be inferred that the thickness of the SiC cladding should be determined by finding the optimal balance between safety during steady-state and safety during accidents.

SiC swelling is a potential source that could introduce considerable uncertainties in SiC cladding structural analysis. While CVD-SiC swelling behavior under irradiation with different temperature is relatively well understood to a level that can be readily used for fuel performance analysis purpose [13], scarce amount of experimental data exists for relevant SiC/SiC composite swelling behavior. This study uses the best estimation of Hi-Nicalon™ Type-S CVI-SiC swelling behavior extrapolated from CVD-SiC behavior as discussed in Section 4.2. A parametric study was conducted to investigate possible departures from the reference SiC/SiC composite swelling behavior. Up to ±20% changes from the reference SiC/SiC swelling in Eq. (39) were tested and the results are shown in Fig. 9 (x = 1.1 denotes 10% increase in swelling in all principal directions).

It can be inferred from Fig. 9 that relatively small uncertainties in the composite swelling can introduce significant stress level variations. A reduction of the composite swelling below the reference level decreases the inner monolith stress level at the expense of putting the composite in stronger tensile stresses. Note that more than 15% reduction in the composite swelling from the reference value starts putting the inner monolith under compressive stresses. Generally, trading off the tensile stress levels of the monolith and the composite through the composite swelling decrease is preferable; it is the norm of the material usage that the composite is far more tolerable of tensile stress than the monolith. Swelling values in the presented reference SiC/SiC composite are very close to ones adopted by a previous work for single-layered SiC analysis [18] as both are basically relying on the swelling values from the same reference [13]. That is the swelling magnitudes of the SiC/SiC are considered basically very close to CVD-SiC. A recent paper [46] by Katoh et al. reviews extensive SiC/SiC composite swelling data as well as those of CVD-SiC. In the paper, they conclude that no convincing reason can be found to believe any appreciable difference between nuclear grade CVD-SiC and SiC/SiC composite irradiation swelling magnitudes. The following statements are quoted from the review of Katoh et al. [46],

“The swelling trends for the CVI composites and the CVD SiC agree well, although the composite swelling in general appears slightly less than the monolithic CVD. However, considering the unfortunately large errors associated with the composite swelling measurement in the present and previous works, this difference may be considered insignificant. There is no significant difference noted between the HNLS and the SA3 fiber composites. Hegeman et al. [47] reports almost identical longitudinal linear swellings for CVD SiC and SA-Tyrannohex™, which is a uni-directional bonded SA3 fiber composite, irradiated side by side in the High Flux Reactor (HFR, Petten) at 600 C and 900 C to 1.1–2.0 dpa. Katoh [48] reports swelling for SA-Tyrannohex close to typical swelling for CVD SiC after irradiation in HFIR at 320 °C to 6 dpa. These two reports imply no significant difference in swelling behavior between the high-purity CVD SiC and the SA3 fiber. Collectively considering these results, there is no convincing reason to believe that the nuclear-grade SiC/SiC composites undergo swelling to an extent significantly different from that of CVD SiC.”

In the research conducted by Newsome et al. [43], their scarce data points to a lower swelling value for SiC/SiC composite at low temperature ~300 °C. For HNLS CVI composite, Newsom’s data [43] give saturated swelling percentages of 1.4% and 0.8% for the irradiated temperature of 300 °C and 800 °C, respectively. While no appreciable difference is found between the CVD-SiC and SiC/SiC composite swelling at high temperature ~800 °C, their difference is found to be noticeable at low temperature ~300 °C; that is, the HNLS CVI composite exhibits approximately 0.5% lower swelling percentage at 300 °C. 0.5% volumetric swelling percentage implies ~0.17 isotropic strain in principal directions, which is considerable amount of strain for ceramic materials. Yet, among the two data points at the irradiated temperature of 300 °C of Newsome’s study [43], one data point exhibits an error bar spanning approximately from 1.1% to 2.4% in volumetric swelling. Such an uncertainty in swelling percentage is intolerably big to structural modeling. In the same manner, as shown in Fig. 9, even 10% departure from the reference swelling value affects stress distributions noticeably. At irradiated temperature 330 °C, ±10%
and SiCf/SiC composite was reported [46]. In Fig. 10, use of thermal expansion coefficient $4.66 \times 10^{-6}$ (1/K) for both CVD-SiC and SiCf/SiC composite was suggested by Katoh et al. [46] at 721 K is compared with that of the reference case, $4.6 \times 10^{-6}$ and $4.0 \times 10^{-6}$ for CVD-SiC, and SiCf/SiC composite, respectively. While the general trend of stress distribution stays unchanged, use of the equal thermal expansion coefficient for both CVD-SiC and SiCf/SiC leads to a departure from the reference case. It slightly increases the tensile stress level in the monolith and reduces it in the composite, primarily by the increased CTE of the composite ($4.0 \times 10^{-6} \rightarrow 4.6 \times 10^{-6}$) while meeting the strain compatibility at the interface. The cladding fracture probability, in this case, is anticipated to increase because of the higher stress level in the monolith.

Some degree of pre-stressing is anticipated to be present in the inner monolith in the course of the fiber winding. This pre-compression can be simply modeled by having a constant compressive mechanical stress added to the total stress. Yet, the magnitude of pre-compression is anticipated to be negligible compared to the three major stress sources found in this analysis.

Failure probabilities for the relevant SiC cladding fracture modes for the burnup dependent reference cladding designs and the relevant departures from the reference cases at high burnup (EOL) conditions are summarized in Table 6.

For the reference BOL case, the cladding is anticipated to reliably carry loads arising from the very early phase of the reactor operation. Yet, SiCf/SiC composite matrix cracking may take place in a number of fuel rods. Matrix cracking could potentially cause EBC cracking through cracking of the CVI overcoat. In a laminated structure, a local fracture of EBC would not necessarily lead to a global failure of the layer. Yet, global composite matrix cracking could be of particular concern for potential strength degradation of the SiC composite due to H2O–C reaction in the presence of cracks in the EBC. In case of the composite swelling magnitudes reported by Newsome et al., fracture of SiC cladding is anticipated to occur by strength failure of the SiCf/SiC layer at EOL while the inner monolith stays under compression. Yet, note that the cladding load failure probability (0.3609) in this case is lower than the reference case (0.9997) at EOL because the lower composite swelling magnitudes of Newsome’s data [43] allocate some added tensile stresses in the composite but leaves the inner monolith under compressive stresses.

For the conservative reference case, it is clear that the structural integrity of the inner CVD-SiC monolith under tensile stresses due to swelling is of prime concern. Adherence to the current triple-layered concept with no changes in material properties and operating conditions can limit the feasibility of the SiC cladding concept. The following discussion addresses possible material property advancements, operating conditions, and cladding...
designs that could help increase structural reliability of SiC cladding for LWRs.

6. Options for advancement of SiC cladding structural integrity: material, core design, and cladding design options

In addition to understanding the structural integrity of the reference cases, it is important to investigate relevant options for improving SiC cladding structural integrity. Such ways can be generally categorized into (a) material property improvements, and (2) changes of reactor design. In addition, we propose a novel duplex SiC cladding design that employs the SiC/SiC composite and the CVD-SiC, as the inner and the outer layer, respectively. Fig. 11 shows cladding load failure probabilities for the considered options with respect to increasing fuel internal pressure levels. Considering that the SiC cladding properties and swelling quickly saturate after a very short irradiation exposure time (~0.01–0.1 DPA), the fuel rod internal pressure can be directly related to burnup (or incore residence time).

The SiC/SiC composite has room for improvements through engineering fiber-architecture. Among the potential advancement of key material properties, improvement of the composite thermal conductivity would affect the stresses arising from thermal expansion and temperature-dependent swelling by inducing a ‘flatter’ radial temperature distribution. The reduced cladding stress level with the composite thermal conductivity increase (1.5 W/mK → 3.0 W/mK or 4.5 W/mK) significantly reduces failure probability up to a certain internal pressurization. The reference composite thermal conductivity of 1.5 W/mK was measured on composites having a multi-layer interphase (sequential pyrolytic carbon and SiC layers). The conductivities of the these composites are approximately half of SiC composites having single-layer interface [46]. Hence, the composite thermal conductivities of 3.0 W/mK and 4.5 W/mK can be regarded for the single-layered SiC/SiC composite.

Manufacturing more statistically reliable CVD-SiC, reflected in the Weibull modulus increase (m_{mod} = 7.5 → 8.5) helps reduce cladding load failure probabilities. Linear fuel power of 18.0 kW/m is an average value of the typical PWR and 25% and 50% reduction of the fuel pin power (13.5 kW/m, and 9 kW/m, respectively), gives a few orders of magnitude reductions in the failure probabilities, with respect to the reference case as shown in Fig. 11. Fuel pin power determines thermal and swelling induced stress levels; a higher fuel pin power induces a steeper temperature gradient, resulting in an overall higher tensile stress in the inner monolith due to predominant swelling. The substantially decreased cladding failure probability with decrease of the fuel pin power implies that SiC cladding has greater potential to be applied to low power nuclear reactors, such as small modular reactors (SMRs). Additionally, the short fuel pin with 50% decrease in its height in some SMRs slightly reduces cladding load failure probability by the weakest link theory in Weibull statistics. Yet, it is inferable in Fig. 11 that the design options that lower stress levels have a higher impact on reducing cladding load failure probabilities than the options alleviating the statistical nature of brittle fractures represented in the Weibull distribution. In Fig. 11, it is important to note that the failure probability variations become smaller with increasing fuel rod internal pressure. This is because cladding failure predominantly occurs by tensile stresses in the inner monolith with a sufficient internal pressurization.

A significant reduction in fuel pin power and improvement of the composite thermal conductivity (or combinations of those two) are found to be potentially the most effective way of advancing SiC structural integrity in steady-state operation. 200% of the composite thermal conductivity increase and 50% of the pin power decrease have almost the equal effect in terms of lowering SiC cladding failure probability (see Fig. 11). Yet, those material improvement and core design changes are not considered to be near-term options; material advancement generally requires a considerable development time and fuel pin power decrease calls for a significant departure from the current LWR designs. For the equal CTE case = 4.66 × 10^{-6} (1/K) for both CVD-SiC and SiC/SiC from Katoh et al. [46] – the increased CTE of the SiC/SiC from 4.0 × 10^{-6} to 4.66 × 10^{-6} slightly increases stress level in the monolith. Because of the low Weibull modulus for monolithic ceramic structure, a slight increase in the tensile stress level sensitively increases failure probability. Possibility of reduced values for the composite swelling magnitudes given by Newsome’s study [43] is also considered. In such a case, strength failure of the cladding occurs primarily by the tensile stress in SiC/SiC layer as discussed in Fig. 10. In the reference cladding design, use of the smaller composite swelling magnitude predicts lower fuel failure probability with increasing internal pressure because of tensile stress tolerance of the composite. However, even with those changes in fuel design and swelling magnitudes, SiC structural integrity is anticipated to be seriously challenged under high burnup operating conditions towards EOL as the fuel rod internal pressure approaching to 18 MPa gives cladding load failure probability on the order of ~10^{-1}.

This illuminates that, with the current triple-layered SiC cladding design, the anticipated benefits of the considered key structural improvements noticeably diminish with burnup. Such vulnerability of the triple-layered cladding structural integrity with increasing burnup is because the inner monolithic CVD-SiC has to bear tensile stresses coming from increasing fuel internal pressurization added to the strong tensile stresses caused by the swelling; the reported cladding load failure probabilities of are dictated by those of the inner CVD-SiC monolith. Consequently, it can be concluded the triple-layered SiC cladding does not effectively utilize unique features of its constituent layers; tensile loads should be predominantly carried by the SiC/SiC composite while the CVD-SiC are present for hermiticity and water corrosion resistance.

In this study, we considered a duplex SiC cladding that utilizes the SiC/SiC_{2} composite as the inner layer, bearing tensile stresses, and the CVD-SiC monolith for the outer layer bearing little tensile or compressive stresses while maintaining hermeticity and resistance to chemical reaction with water, as illustrated in Fig. 12. Such a structure can be manufactured by winding the SiC fibers around a central graphite rod, which is later selectively removed by burning at high temperature.

The reference duplex design takes the same total thickness as the typical Zr cladding (0.57 mm), with relative layer fraction of
1:1 for the inner composite and the outer CVD-SiC monolith. The stress distributions of the proposed Duplex SiC cladding are compared with those of the reference triple-layered cladding at BOL and EOL conditions as shown in Fig. 13.

At BOL, the duplex SiC cladding is found to be under compression. At EOL, the highest tensile stress occurring at the innermost surface is now experienced by the composite. Comparing with the tensile stress levels for the inner CVD-SiC monolith of the triple-layered design at EOL, the outer CVD-SiC monolith in the Duplex sees much lower tensile stresses as shown in Fig. 13. Hence, the proposed Duplex design properly allocates peak tensile stresses to the composite while minimizing tensile stress levels in the monolith, which avoids/reduces the structural vulnerability of the monolithic ceramic under tension. As a consequence, the Duplex cladding exhibits much lower cladding load failure probabilities than the reference design of a triple-layered cladding as shown in Fig. 11. Up to an internal fuel rod pressurization of 18 MPa, the cladding load failure is dictated by the outer monolith fracture. Yet, further internal pressurization increases stresses at the inner surface of the SiC cladding and noticeably raises the cladding load failure probability by the ultimate failure of the inner composite (the increasing gap between the duplex cladding load failure and the monolith fracture indicates an increasing ultimate failure probability of the composite in Fig. 11). The proposed Duplex SiC cladding design is worth investigation as an alternative design that could simultaneously improve structural integrity, hermeticity, and chemical resistivity of SiC cladding.

7. Conclusions and recommendations for future work

The fracture of a multi-layered SiC clad of LWR fuel rods during in-core service was investigated analytically. A model that captures the 3D stress fields in each layer in the cladding was developed and validated by comparison to the FEA code, ANSYS-Mechanical. The stress fields resulting from pressure differences, thermal stresses, and swelling were used to develop a statistical fracture model for the SiC cladding. The conditions under which the cladding will operate were assumed typical of a present day PWR in terms of linear heat rate and initial geometry. All layers are assumed to be stress free as they start operation in the reactor, which ignores any preexisting compressive stresses in the innermost layer due to the winding of the fibers around it.

In steady-state operations, stress distributions are subject to a noticeable change with incore residence time. During the low burnup operation (BOL), the SiC cladding is under compressive stresses for most of its radial regions because of strong compressive mechanical stress and absence of swelling induced stresses. After a short period of radiation exposure until the high burnup operation (EOL), irradiation swelling and further internal pressurization put the inner monolith – the region of high temperature – in tension, despite of the positive effects of compressive thermal stresses. As a consequence, structural integrity of the inner CVD-SiC monolith is a limiting factor of a triple-layered SiC cladding, which may seriously affects feasibility of the LWR SiC cladding concept. However, the results are sensitive to the assumed SiC swelling and thermal conductivity, and thus swelling and thermal conductivity are sources of considerable uncertainties in SiC cladding structural analysis. Further evaluation of the SiCf/SiC composite swelling behavior and effective thermal conductivity will clarify feasibility and design directions of the SiC cladding concept for LWRs.

Developing a SiC fiber-reinforced composite with higher conductivity (100–200% increase) would benefit the integrity of the cladding significantly, as the swelling-induced loads arising from differential temperature are considerable fractions of the total stresses. Similarly, SMRs with a significantly reduced pin power (20–50% decrease from the reference PWR) are also considered to be more accommodating of the SiC cladding. However, even with those changes, SiC structural integrity is anticipated to be seriously challenged at high burnup operating conditions due to considerable internal pressurization; the fuel rod internal pressure approaching to 18 MPa gives cladding load failure probability on the order of ~10−3. The fission gas release is affected by the fuel temperature and thus, designs that lower the fuel temperature would be needed. This illuminates limitations of the widely-recognized triple-layered SiC cladding design.

In this study, a double-layered SiC cladding that employs the SiCf/SiC composite, and the CVD-SiC for the inner, and outer layer, respectively, has been analyzed. This cladding design exhibits a significantly reduced SiC cladding failure probability under steady-state operating conditions as it appropriately allocates peak tensile stresses in the inner composite while significantly reducing tensile stress levels of the CVD-SiC monolith – a direction in compliance with the norm of the material usages of the composite and the monolithic ceramics.

Development of more statistically consistent Weibull parameters for both the composite and CVD-SiC, with better understanding of the strength dependence on size in the laminated form,
would increase confidence in the predictions by the statistical model of their fractures. In that regard, experimental studies on the statistical strength failure of SiC cladding under representative operating condition are necessary at some point in the future. In addition, some modeling efforts should be made to increase precision of the current statistical approach in an extremely low failure rate. That is, validity of the weakest link theory for individual components of the SiC cladding in a laminated form should be further investigated. This issue boils down to – how to properly extrapolate the strength, and fracture behavior of the lab scale specimen to the actual cladding that exists in a laminated form. This study treated the fiber-reinforced composite as another layer of the SiC cladding. As a result, the fiber-reinforced composite was mechanically treated in the same manner as the monolith using representative composite properties. However, mechanical behavior of a fiber-reinforced composite involves more complex mechanisms, as it involves interactions between fibers, and between the fibers and matrices. Orientation of fibers and the influence of the direction of woven fiber fabrics on the mechanical behavior of the composite should be further analyzed. In addition, a slightly different thermal expansion coefficient of the fibers and the matrix causes additional local stresses. Indeed, mechanical behavior of fiber reinforced composites with respect to its fiber architecture has been extensively studied in consideration of various applications including aerospace application, and combustion engines, etc. Yet, very limited literature is available for the fiber reinforced composite specific to SiC cladding applications in LWRs. Although the reported high Weibull modulus of the composite leads to reduced ‘volume sensitivity for the material strength’ with the weakest link fracture assumption, a more fundamental investigation of the composite statistical fracture with respect to its size as a cladding material needs be conducted. This theoretical study on SiC cladding fracture, nevertheless, provides an important tool and foundation to predicting SiC cladding fracture behavior in both steady-state and accident conditions.

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Appendix A

Stress distributions for Duplex cladding obtained by the code developed in this study were validated by comparing results independently obtained by FEA solution with ANSYS-Mechanical. The same procedures discussed in Section 4.3 were used. The developed model gives a good agreement with FEA solutions for the Duplex stress distributions as shown in Fig. A.1.

Appendix B

Although no information is certain about the observed cracking due to the poor nature of the mount used, the cracks observed in the inner monolith of a triple-layered SiC fuel that reached 20 MWd/kgU in the simulated LWR pin power and irradiation conditions at HFIR in Oakridge National Laboratory [49] (Fig. B.1) could be potential evidence for swelling induced stresses in the inner monolith.

Fig. A.1. ANSYS-mechanical validation for a duplex SiC cladding design in the reference high burnup conditions (EOL). Inner composite: outer monolith = 4.5:5.5.

Fig. B.1. Triple-layered SiC cladding sample from incore fuel pin test at HFIR in oakridge national laboratory after 20 MWd/kgU [42].

References
