Design and material issues for high performance SiC$_f$/SiC-based fusion power cores

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Abstract

The SiC$_f$/SiC composite is a promising structural material candidate for fusion power cores and has been considered internationally in several power plant studies. It offers safety advantages arising from its low induced radioactivity and afterheat, and the possibility of high performance through high temperature operation. However, its behavior and performance at high temperatures and under irradiation are still not well known and need to be better characterized. This paper summarizes the current SiC$_f$/SiC design and R&D status. The latest SiC$_f$/SiC-based power core design studies are summarized, and the key SiC$_f$/SiC parameters affecting the performance of power core components are highlighted. The current status of the material R&D is discussed, with the focus on fabrication and joining, baseline properties and properties under irradiation, as well as the desirable evolution of these properties. In the light of this, the R&D plans are summarized and assessed. Finally, to help present-day design studies and in the expectation of future confirmatory R&D results, recommendations are provided on SiC$_f$/SiC parameters and properties to be assumed for present design analysis of long term SiC$_f$/SiC-based power plants. © 2001 Elsevier Science B.V. All rights reserved.

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1. Introduction

The use of the SiCf/SiC composite as structural material in a fusion reactor can be viewed as a high-risk high-payoff endeavor. The high payoff is linked to the superior safety characteristics of SiC arising from its low induced radioactivity and afterheat, and to the possibility of high performance through high temperature operation. The high risk is associated mainly with the uncertainty about SiCf/SiC behavior and performance at high temperatures and under irradiation.

SiCf/SiC was considered several years ago in the ARIES-I [1], ARIES-IV [2] and PROMETHEUS [3] power plant studies in the US and more recently in other international studies, in particular the TAURO [4] and the He-cooled pebble bed blanket [5] design studies in the EU, the DREAM [6] reactor study in Japan, and the ARIES-AT [7] study in the US. Several issues have been identified for SiCf/SiC, including the cost of fabrication, joining methods, and factors limiting the range of operation. The major issues appear to be the rather low thermal conductivity at high temperatures and under irradiation, and the maximum allowable operating temperature.

Over the past several years an international R&D effort has helped address and better understand some of these issues. This paper aims to assess the current design and R&D status. First, SiC application in fusion power plants is summarized. Example of the latest fusion power core design concepts utilizing SiCf/SiC composites in the EU, Japan and the US are then briefly described and the key SiCf/SiC issues influencing their performance are discussed. Next, the current status of the material R&D including fabrication and joining, baseline properties and properties under irradiation is described and the extent to which these provide information to help resolve the design issues is assessed. The desirable evolution of SiCf/SiC parameters and properties is discussed with the aim of increasing the performance and attractiveness of SiCf/SiC-based fusion power plant designs. Finally, future R&D plans are outlined.

2. Design concepts

2.1. Use of SiCf/SiC in fusion power plants

Silicon carbide is a refractory semiconductor with unique properties. Because of its thermal, mechanical and chemical stabilities, it can be used in extremely harsh environments. Pure SiC also provides exceptionally low radioactivity under 14-MeV neutron radiation, making it one of the top candidates for use in fusion power cores.

Due to its brittle nature, SiC often has been considered in a fiber-reinforced ceramic matrix composite form. This enables its use as a structural and/or pressure vessel material subjected to high thermal and pressure stresses. For example, SiCf/SiC can be fabricated with sufficient thermal conductivity to serve as a first wall under reasonable heat loads (up to %1 MW/m²). Containment of coolants at high pressure, especially helium, is a concern. However, various techniques are available to coat or impregnate surfaces to provide hermeticity.

Behind the first structural wall, SiC can provide other attractive features. Its strength and chemical stability at high temperatures enables very high coolant outlet temperatures, providing the possibility of very high thermal conversion efficiency. By using SiC within the pressure boundary, instead of as the pressure boundary, its high temperature capability can be fully exploited. For example, blanket designs have been considered in which low-conductivity SiC is used as a thermal insulator to allow the pressure vessel to remain at a lower temperature than the coolant outlet.

If SiC is used to contain the main coolant within the power core, especially if the coolant temperature is raised to extremely high values, then SiC is probably needed in the entire primary cooling circuit. SiC pipes can be manufactured at reasonable cost; however, high-performance heat exchangers made from SiC may be complex and expensive.

SiC has some other interesting speciality uses for some blanket concepts. With its modest electrical conductivity (of the order of 500 Ω m⁻¹, and probably lower for composites because of the fiber/matrix interfaces), SiC can be used with
some liquid metals to reduce the MHD body force. For example, it has been proposed as a ‘flow insert’ for this purpose. SiC also has very low hydrogen permeability, making it a candidate for tritium permeation barriers.

In regions of very high heat flux, such as a tokamak divertor, SiC is not particularly well-suited. If a separate plasma-facing armor material is needed, then issues related to so-called ‘duplex structures’ arise. Bonding methods exist, but special care must be taken. One especially attractive duplex system is SiC with tungsten, since both share similar thermal expansion coefficients and both can operate at very high temperatures.

While there are many potential applications of SiC and SiCf/SiC composites in fusion power plant designs, special attention is given in this article to ‘high performance’ applications in a fusion power core that place the most severe requirements on the material. Designs studies are useful tools to define which parameters are most critical in order to establish the feasibility and to fully exploit the advantages of SiC/SiC composites for such applications. The requirements on the materials can be established through detailed analysis of the thermal, mechanical and nuclear performances of specific design concepts. As highlighted in the latest design studies summarized below, the parameters that emerge as the most important for the success of a SiC/SiC-based power core include:

- thermal conductivity of the SiC/SiC composite at high temperatures and under irradiation;
- the maximum SiC/SiC temperature limit under irradiation;
- the maximum Pb–17Li/SiC interface temperature under flowing conditions, including the effect of irradiation, if any;
- maximum stress limits to be used for SiC/SiC;
- fabrication and joining (brazing) of SiCf/SiC and bonding of W to SiCf/SiC, including methods and constraints such as minimum thickness;
- lifetime.

2.2. TAURO breeder blanket concept

The TAURO blanket is a self-cooled Pb–17Li blanket using SiCf/SiC structures [8]. It has been developed with the objective of reaching passive safety in tokamak-type Fusion Power Reactors. The TAURO design minimizes the energy available in the vacuum vessel in order to avoid breaking of the confinement, which would lead to the release of radioactive materials to the environment. This requirement leads to the use of low-pressure fluids with low reactivity and low afterheat levels. Liquid metal coolants appear to be the best choice, especially the eutectic Pb–17Li.

The TAURO blanket offers the capability of heat extraction at high coolant temperatures and promises favorable conversion efficiencies. Recently performed thermo-mechanical analyses have shown that the blanket could withstand a surface heat flux of 0.7 MW/m². In the reference design, which assumes a surface heat flux of 0.5 MW/m², Pb–17Li inlet/outlet temperatures are, respectively, 450 and 860°C and the estimated efficiency is approximately 47%. The possibility of assuming a higher thermal conductivity than the 15 W/m K originally used in the analysis implies increases of both the surface heat flux and the conversion cycle efficiency without exceeding the maximum temperature and the stress limits for SiC/SiC.

The configuration and the performance of the TAURO blanket design are described below, along with a discussion on the improvements required in the characteristics of present-day SiCf/SiC composites, in particular its thermal conductivity, neutron irradiation resistance and compatibility with Pb–17Li at high temperature [9,10].

2.2.1. Specifications and design description

The design is based on the specifications for the SEAFL study [11], i.e. fusion power of 3000 MW, neutron wall load of 2 MW/m² and surface heat flux of 0.5 MW/m². The TAURO blanket is essentially formed by a SiCf/SiC box with an indirectly-cooled first wall, which acts as a container for the Pb–17Li. The Pb–17Li acts as a coolant, a breeder multiplier material, and a tritium carrier. Each outboard segment is poloidally divided into several straight modules, attached on one common thick back-plate. The number of modules cooled in series depends on the assumption on inlet and outlet Pb–17Li temperatures. The feeding pipes are lo-
cated behind the module. The coolant enters the inlet collector through a single tube and is divided into five sub-flows, one for each sub-module (see Fig. 1). Within each submodule the Pb–17Li flows at first poloidally downward in a thin channel located just behind the first wall, makes a U-turn at the bottom into a second channel and flows up, then down into the outlet collector. Toroidal plates (stiffeners) are required for reinforcing the sub-module box against the hydrostatic pressure (assumed to be 1.5 MPa in all calculations) and act as flow separators [12].

2.2.2. Design criteria and models

The design optimization and structural assessment of the TAURO blanket were performed in 3D, assuming square meshes and an orthotropic model for SiCf/SiC. Due account was taken of the poloidal variations in heat transfer, coolant conditions, mechanical loads and thermal loads. Conduction remains the dominant heat transfer mechanism; the thermal conductivity of SiCf/SiC was assumed to be 15 W/m K in all directions. The other assumed SiCf/SiC characteristics were those of the industrial Cerasep® N3-1 [9].

The TAURO design criteria [13], discussed in Section 3.3, have been applied. The assumed limits are the following:
- the tensile and compressive stresses in plane are limited to 145 and 580 MPa, respectively;
- the tensile and compressive stresses through the thickness are limited to 110 and 420 MPa, respectively;
- the shear stresses through the thickness are limited to 45 MPa.

These limits have been determined using experimental results obtained several years ago on Cerasep® N2-1 and are, therefore, probably too pessimistic when applied to present-day SiCf/SiC or to future improved composites.

The following constraints have also been used:
- the maximum SiCf/SiC temperature $<1300$°C (expected for future composites based on Hi-Nicalon fibers);
- the average coolant outlet temperature $>850$°C to achieve a conversion efficiency $>45\%$.

The maximum SiCf/SiC temperature could be lower due to the void swelling regime, but no data are available to determine the temperature transition from the dislocation loop to the void swelling regime. The maximum temperature may be around 1000°C because of this transition.

2.2.3. Neutronic analysis

A 2D neutronic analysis of the TAURO blanket has been performed with the Monte-Carlo code TRIPOLI4 using the ENDFB-VI transport cross section library. The adopted model reproduces half sector of the blanket geometry at the equatorial plane. It results in a TBR of 1.37 is
Table 1
Cerasep® N3-1 main characteristics

<table>
<thead>
<tr>
<th>Property</th>
<th>Temp. (°C)</th>
<th>Measured value (SEP data)</th>
<th>Value assumed for the analyses</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density (g/cm³)</td>
<td>20</td>
<td>&gt;2.4</td>
<td>2.5</td>
</tr>
<tr>
<td>Porosity (%)</td>
<td>20</td>
<td>10 ± 2</td>
<td>10</td>
</tr>
<tr>
<td>Fiber content (%)</td>
<td>20</td>
<td>40</td>
<td>40</td>
</tr>
<tr>
<td>Thickness (mm)</td>
<td>–</td>
<td>0.8–6</td>
<td>6–10</td>
</tr>
<tr>
<td>Tensile stress (in plane) (MPa)</td>
<td>20</td>
<td>(300 ± 20)</td>
<td>–</td>
</tr>
<tr>
<td>Tensile strain (%)</td>
<td>20</td>
<td>0.80 ± 0.25</td>
<td>–</td>
</tr>
<tr>
<td>Trans-laminar shear stress (MPa)</td>
<td>20</td>
<td>200 ± 20</td>
<td>–</td>
</tr>
<tr>
<td>Inter-laminar shear stress (MPa)</td>
<td>20</td>
<td>44</td>
<td>44</td>
</tr>
<tr>
<td>Young’s modulus (in plane) (GPa)</td>
<td>20</td>
<td>200 ± 20</td>
<td>200 a</td>
</tr>
<tr>
<td>Shear modulus (in plane) (GPa)</td>
<td>20</td>
<td>–</td>
<td>80 b</td>
</tr>
<tr>
<td>Shear modulus (through thickness) (GPa)</td>
<td>20</td>
<td>–</td>
<td>50 b</td>
</tr>
<tr>
<td>Poisson’s ratio (in plane)</td>
<td>20</td>
<td>–</td>
<td>0.18 a,b</td>
</tr>
<tr>
<td>Poisson’s ratio (through thickness)</td>
<td>20</td>
<td>–</td>
<td>0.18 a,b</td>
</tr>
<tr>
<td>Thermal conductivity (in plane) (W/m K)</td>
<td>1000</td>
<td>15</td>
<td>15</td>
</tr>
<tr>
<td>Thermal conductivity (through thickness) (W/m K)</td>
<td>800</td>
<td>7.6</td>
<td>15</td>
</tr>
<tr>
<td></td>
<td>1000</td>
<td>7.5</td>
<td>15</td>
</tr>
<tr>
<td>Thermal expansion coefficient (in plane) × 10⁻⁶ K⁻¹</td>
<td>20</td>
<td>4</td>
<td>4</td>
</tr>
<tr>
<td>Thermal expansion coefficient (through thickness) × 10⁻⁶ K⁻¹</td>
<td>20</td>
<td>–</td>
<td>2.5 b</td>
</tr>
</tbody>
</table>

a Value not available; same value has been assumed for in-plane and through thicknesses.
b Corresponding value for the 2D composite.

obtained with 90% enrichment. A rough 3D estimation would give a TBR of 1.17. Also, a detailed distribution of the power density deposition has been obtained that will serve as a source in the thermo-mechanical calculations.

2.2.4. Thermo-mechanical analyses
This section summarizes the main results of the recent thermo-mechanical analyses, which focused on the outboard blanket in the mid-plane region where the thermal conditions are the most severe. The analyses were performed with CASTEM2000 [14,15] using the properties of Cerasep® N3-1 shown in Table 1. The evaluated maximum surface heat flux has been assumed along the whole sub-module height. The results are obtained by using a non-linear mechanical model that simulates the non-linear stress-strain relation and the TAURO design criteria presented in Section 3.4. The Von Mises stress is also reported for comparison with previous analyses.

Assuming two modules cooled in series, a surface heat flux of 0.5 MW/m², an inlet temperature of 750°C, a Pb–17Li velocity in the first layer of 1.3 m/s, a first wall thickness of 6 mm, a module width of 0.3 m and a module height of 2 m, the coolant outlet temperature is around 860°C. The highest stresses appear on the first wall. The TAURO criterion for plane stresses is 0.93 (below the limit of 1), the maximum normal stress through the first wall thickness is 63 MPa, and the maximum shear stress is 30 MPa. The maximum tensile stress in plane (evaluated with the Von Mises criterion) is 125 MPa. The module expands about 3 mm in the poloidal direction. The maximum SiCf/SiC temperature is about 1050°C.

2.2.5. TAURO blanket thermo-mechanical limits
An estimation has been performed of the TAURO blanket performance with higher surface heat flux and correspondingly higher neutron wall load. Increased power densities are
expected to yield reduced electricity costs and a more economically attractive power reactor. Figs. 2 and 3 show the evolution of maximum temperature and stress as the power deposition increases. For a module height of 2 m, a first wall thickness of 6 mm, and an inlet temperature of 750°C, the maximum temperature design limit is reached at 0.69 MW/m², corresponding to a neutron wall load of 2.8 MW/m². The maximum tensile stress in plane in the first wall is reached at a surface heat flux of 0.80 MW/m².

The reduction of the first wall thickness, and of the module height promote higher surface heat flux and a neutron wall load of 3–4 MW/m² can be envisaged (see Table 2). This evaluation gives an idea of the sensitivity of the design to possible improvements of the calculation models and/or to an increase of the acceptable stress limits. For the TAURO blanket, the coolant temperature does not affect the stress limit when increasing the surface heat flux. This suggests that, under the conditions considered here, high power densities are, in principle, favorable for higher conversion efficiencies [4].

2.2.6. Impact of thermal conductivity

The thermal conductivity is the major issue for this structural material because it affects the blanket temperature distribution (maximum SiC$_f$/SiC temperature and outlet Pb–17Li temperature), and the cycle efficiency. If the transverse thermal conductivity is increased to 20 W/m K, the maximum acceptable surface heat flux for the TAURO blanket would increase by approximately 12%. The increased demand for improvement of the SiC$_f$/SiC characteristics suggests that a value of the surface heat flux of 0.7–0.8 MW/m² is probably a compromise. On the other hand, a higher thermal conductivity promotes the heat exchange between the first wall and the Pb–17Li, and reduces the maximum SiC$_f$/SiC temperature. The double consequences are the possibility to increase the height of the

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**Fig. 2.** Maximum temperature in the TAURO first wall as a function of surface heat flux for different first wall thicknesses, $\delta$. The Pb–17Li temperature rise as a function of surface heat flux is also shown. (Module height = 2, 0.3 m; Pb–17Li inlet temperature, pressure, velocity = 750°C, 1.5 MPa, 1.3 m/s; SiC$_f$/SiC thermal conductivity = 15 W/m K).

**Fig. 3.** Maximum stress in the TAURO first wall as a function of surface heat flux for different first wall thicknesses, $\delta$. (Module height, width = 2, 0.3 m; Pb–17Li inlet temperature, pressure, velocity = 750°C, 1.5 MPa, 1.3 m/s; SiC$_f$/SiC thermal conductivity = 15 W/m K).
Table 2
Performances of the TAURO blanket for different module lengths and first wall thicknesses

<table>
<thead>
<tr>
<th>Module length (m)</th>
<th>2</th>
<th>1</th>
</tr>
</thead>
<tbody>
<tr>
<td>First wall thickness (mm)</td>
<td>6</td>
<td>3</td>
</tr>
<tr>
<td>Maximum allowable surface heat flux based on stress limit (MW/m²)</td>
<td>0.88</td>
<td>0.98</td>
</tr>
<tr>
<td>Maximum allowable surface heat flux based on temperature limit (MW/m²)</td>
<td>0.69</td>
<td>0.98</td>
</tr>
<tr>
<td>Corresponding neutron wall load (MW/m²)</td>
<td>2.8</td>
<td>3.9</td>
</tr>
</tbody>
</table>

* With Pb–17Li inlet temperature, pressure and velocity of 750°C, 1.5 MPa and 1.3 m/s, respectively.

module and the outlet temperature of the Pb–17Li within the stress limits for SiC/β-SiC. An extreme case — assuming a thermal conductivity of 20 W/m K, a thickness of 6 mm, and a height of 3 m — has showed the possibility to increase the outlet temperature up to 940°C without exceeding the maximum temperature and the stress limits for SiC/β-SiC. The conversion efficiency could be increased correspondingly to 50%.

2.2.7. External circuits and components

The TAURO blanket could be associated with an external primary circuit for power conversion and tritium recovery. The primary circuit consists of several parallel circuits, each of them conveying about 600 MW of thermal power. Its principal components are a heat exchanger, a tritium extraction unit and, of course, a Pb–17Li pump. The tritium extraction unit could be a gas/Pb–17Li contactor in a by-pass configuration, common for all the Pb–17Li circuits and treating a relatively small Pb–17Li flow-rate (depending on the tolerable tritium concentration in the Pb–17Li).

At present, a Brayton cycle using helium is foreseen for power conversion, which eliminates the potential risk of Pb–17Li/water interaction although it requires Pb–17Li/helium heat exchangers, the technology for which is yet to be developed [15,16]. The helium pressure should be high to minimize pressure drop and component sizes. A rupture disk in the Pb–17Li circuit close to the heat exchanger and a discharge vessel will be required to protect the blanket from accidental pressurization in case of heat exchanger failure. When assuming TAURO coolant inlet/outlet temperatures of 450/860°C, the power conversion efficiency would theoretically attain 60%. However, when including the estimated loss due to the various components, the conversion efficiency is 47%.

Such a conversion efficiency could be increased up to 50% if the Pb–17Li inlet temperature is increased to 650°C or the outlet temperature to 1000°C. Therefore, the precise determination of the global blanket system efficiency depends on an optimization between blanket module arrangement, pressure drop, Pb–17Li heat-up and power conversion.

2.3. DREAM power core

Much of the difficulty associated with the maintenance of tokamak systems arises from the configuration complexity and material radioactivity. The DREAM (DRastically EAsy Maintenance) power plant has been conceived to mitigate this difficulty by considering a torus configuration with a high aspect ratio of 8 and a low-activation structural material (SiCf/β-SiC composite) [6,17–19]. The torus system is divided radially into equal sectors and each sector forms an assembling unit. All the piping and feeder systems are extracted to the spacious torus central region. The maximum toroidal field strength of 20 T and the plasma major and minor radii of 16 and 2 m lead to a high fusion power of 5.5 GW under moderate physical constraints such as a non-reversed shear profile with a MHD safety factor of 3, a Troyon factor of 3 for the beta limit and an energy confinement time enhancement factor of 2. The average neutron wall load is 2.5 MW/m². Surface heat loads on the blanket first wall and the diverter plate are estimated to be about 0.5 and 5 MW/m², respectively.
The power core is divided into 16 sectors corresponding to the 16 TF coils. Each sector has its own horizontal maintenance port, allowing replacement of the entire sector without opening the cryostat or disassembling other components such as the coil system. Details of the maintenance procedure can be found in Refs. [6,17–19].

2.3.1. Blanket system

2.3.1.1. Blanket module configuration. Each blanket sector is divided into 16 sub-sectors (rings) in the toroidal direction. The blanket modules are connected by bolts to the cooling rings, which also function as support structures. The toroidal dimension of the blanket modules is kept under 500 mm based on considerations of fabricability and maintainability. Their poloidal and radial dimensions are 500 and 650 mm, respectively.

A typical blanket module is shown in Fig. 4. It consists of a first wall zone, a tritium breeding zone and a high-temperature shielding zone. The wall of the module has cooling paths in it. Neutron multiplier material (Be), tritium breeding material (Li₂O or other lithium ceramics) and shielding material (SiC) are packed in the module as small size pebbles of diameter 1 mm for Be and Li₂O, and 10 mm for SiC. The He coolant supplied from the inlet pipe of the cooling ring first flows through the side wall and first wall channels. It then flows into the module through the porous partition wall and cools the blanket internals before exiting through the outlet pipe of the ring. The first wall region is a 15-mm thick double wall structure consisting of a 4-mm plasma side wall and an 8-mm inner wall between which are located 3 mm × 10 mm rectangular cooling channels.

The blanket module is fabricated by the uniforming method, which is illustrated in Fig. 5 and described in more detail in Refs. [6,17–19]. The SiC/SiC composite must be treated to get a high density. There are two methods to form a ceramic matrix directly from gaseous raw materials: chemical vapor deposition (CVD) and chemical vapor infiltration (CVI). The CVI process is assumed here based on its advantages of no pressurization, close adherence between fiber and matrix, and possibility of complex shaping, although it is recognized that there is less experience in applying this process and that its treatment time is longer.

2.3.1.2. Blanket analysis. A thermo-mechanical analysis of the SiC/SiC first wall during normal operation was performed based on the following parameters:

- nuclear heating rate: 16.5 MW/m³
- surface heat load: 0.5 MW/m²
- heat transfer coefficient: 6000 W/m² K
- blanket He inlet and outlet temperatures: 600 and 900°C
- first wall He average temperature: 700°C
- coolant pressure: 10 MPa

The properties of the SiC/SiC composite used in the analysis are summarized in Table 3. In the absence of confirmed thermal conductivity values of high quality SiC/SiC under irradiation, two values were assumed in the analysis: 15 W/m K as a reference case and 60 W/m K as an optimistic one. The temperature profiles and stress distributions in the first wall are shown in Fig. 6(A) and (B) for thermal conductivities of 15 and 60 W/
m K, respectively. The corresponding maximum temperatures are 954 and 823°C, which are lower than the operating temperature limit (1100°C) assumed in this study. The stress distributions are shown as the Tresca stresses due to the internal pressure (10 MPa) and the temperature gradients. The maximum values are 137 and 75.4 MPa for thermal conductivities of 15 and 60 W/m K, respectively. These values are also well below the allowable stress limit (200 MPa) assumed in this study.

The pressure drop in the pebble bed blanket region was estimated using Ergun’s equation and the film temperature difference between the helium coolant and the Li$_2$O pebbles was estimated by Shirai’s equation [6,17–19]. Connecting more than one blanket module in series results in a higher He velocity and somewhat better heat transfer. However, the corresponding pressure drop becomes unacceptably high and fully parallel flow between blanket modules was selected for the design. The following limits were used to select the design parameters: the maximum SiC$_f$/SiC temperature at the plasma side $T_a < 1100°C$, the maximum temperature of the breeder pebble $T_b < 1000°C$ and the net thermal efficiency $\eta > 45\%$. The design window under the reference design parameters is shown in Fig. 7.

2.3.2. Divertor system

2.3.2.1. Divertor configuration. The divertor system consists of two principal parts: the target plates and the structures. The target plates include three pieces: inner, outer and ‘dome’ plates. The plasma flows through the scrape-off layer and enters the divertor where enhanced line radiation from injected neon impurity allows much of the power to be distributed along the plates and also

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Specification</th>
</tr>
</thead>
<tbody>
<tr>
<td>Thermal conductivity</td>
<td>Reference case 15 W/m-K</td>
</tr>
<tr>
<td></td>
<td>Optimistic case 60 W/m-K</td>
</tr>
<tr>
<td>Specific heat</td>
<td>1200 J/kg-K</td>
</tr>
<tr>
<td>Density</td>
<td>2500 kg/m$^3$</td>
</tr>
<tr>
<td>Young’s modulus</td>
<td>300 GPa</td>
</tr>
<tr>
<td>Poisson’s ratio</td>
<td>0.2</td>
</tr>
<tr>
<td>Tensile strength</td>
<td>300 MPa</td>
</tr>
<tr>
<td>Allowable stress ($= 2/3 \times$ ultimate strength)</td>
<td>200 MPa</td>
</tr>
<tr>
<td>Thermal expansion coefficient</td>
<td>$3.3 \times 10^{-6}$ K$^{-1}$</td>
</tr>
<tr>
<td>Maximum allowable temperature</td>
<td>1100°C</td>
</tr>
</tbody>
</table>
Fig. 6. Temperature profiles and stress distributions in DREAM blanket first W wall for SiC/SiC thermal conductivities of: (A) 15 W/m K; and (B) 60 W/m K.

partially redirected out to the first wall. A permeation wall is provided to prevent a backflow of neutral gas and impurity into the main plasma. It also promotes recycling of neutral particles. The dome is installed near the X-point along the scrape-off-layer (SOL). It prevents impurity particles generated at the divertor plate from invading the X-point zone. The baffle plate is provided at the inlet of the divertor area, i.e., below the blanket. It prevents a backflow of neutral particles into the main plasma. The baffle plate surface is coated with 2-mm thick tungsten (W) and the divertor plate surface with 2-mm thick CVD-SiC.

Fig. 8 shows the overall divertor configuration. The major design parameters of the divertor system are listed in Table 4. The He coolant is fed through a double-walled pipe. Comb-shaped exhaust ports are installed, facing the inner and outer strike points and work as the permeation wall. Coolant paths are provided in the teeth of the comb. The surface of the divertor plates is coated with 2-mm thick tungsten for the baffle plate and with 2-mm thick CVD-SiC for the divertor. Small 3 mm × 3 mm coolant channels are provided at the front to accommodate the surface heat flux. Larger channels are used at the back where the heat load is lower in order to

Fig. 7. Design window for DREAM blanket first wall.
minimize the pressure losses. The divertor He inlet/outlet temperatures are 600/800°C to maintain the maximum SiC temperature <1100°C. The coolant is then directed to the high-temperature shield, where its temperature is raised to 900°C.

3.2.2. Analysis of divertor plate. A heat transfer coefficient of 10 kW/m² K is required to maintain the 2-mm thick SiC wall maximum temperature

Table 4
Major specifications of DREAM divertor system

<table>
<thead>
<tr>
<th>Items</th>
<th>Design parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Type</strong></td>
<td>Slot-type gaseous divertor</td>
</tr>
<tr>
<td>Max. surface heat load:</td>
<td></td>
</tr>
<tr>
<td>Reference case</td>
<td>5 MW/m²</td>
</tr>
<tr>
<td>Optimistic case</td>
<td>3 MW/m²</td>
</tr>
<tr>
<td>Incident particle energy</td>
<td>~30 eV</td>
</tr>
<tr>
<td>Structural material</td>
<td>SiC/SiC composite</td>
</tr>
<tr>
<td>Thermal conductivity</td>
<td>60 W/m K</td>
</tr>
<tr>
<td>Allowable stress</td>
<td>200 MPa</td>
</tr>
<tr>
<td>Surface coating material:</td>
<td></td>
</tr>
<tr>
<td>Divertor plate</td>
<td>CVD SiC</td>
</tr>
<tr>
<td>Baffle plate</td>
<td>CVD/LPS W</td>
</tr>
<tr>
<td>Coating thickness</td>
<td>2 mm</td>
</tr>
<tr>
<td>Coolant:</td>
<td>Helium gas</td>
</tr>
<tr>
<td>Pressure</td>
<td>10 MPa</td>
</tr>
<tr>
<td>Inlet/Outlet temperature</td>
<td>600/800°C (Reheated in the high temp. shield to 800°C)</td>
</tr>
<tr>
<td>Maximum coolant velocity</td>
<td>110 m/s</td>
</tr>
<tr>
<td>Total pressure loss</td>
<td>1 MPa</td>
</tr>
<tr>
<td>Maximum wall temperature</td>
<td>1100°C</td>
</tr>
</tbody>
</table>
develop materials having a much higher performance or to reduce the divertor load requirements.

2.3.2.3. Sputtering erosion. An analysis of divertor sputtering erosion was done. Details of the analysis and of its basis can be found in Refs. [6,17–23]. Here, a summary of the major results and observations is provided.

1. The sputtering threshold energy of SiC is about 25 eV. It is not necessary to consider self-sputtering because the threshold energy of self-sputtering is large and the self-sputtering yield for a low-Z material does not exceed unity. The analysis results indicate that the erosion rate of SiC increases rapidly when the incident particle energy exceeds 30 eV, and a practical design is not possible.

2. The threshold energy for W is higher than 145 eV but self-sputtering may be critical as the threshold energy for W self-sputtering is low, 90 eV, and the self-sputtering yield for a high-Z material exceeds unity. The analysis results confirm this, showing very small sputtering erosion of W by D, T and He up to 200 eV. Self-sputtering is the critical limiting condition (89 eV) in this case.

In the DREAM design, the following divertor wall design was adopted based on the above observations:

- incident particle energy, \( E_i < 30 \text{ eV} \): SiC coated wall
- incident particle energy, \( 30 \text{ eV} < E_i < 89 \text{ eV} \): SiC wall coated with W
- incident particle energy, \( E_i > 89 \text{ eV} \): no practical design solution found

Fig. 9. Temperature profiles and stress distributions in DREAM divertor plasma-facing wall for surface heat fluxes of: (A) 3 MW/m²; and (B) 5 MW/m².
Though the divertor plasma conditions are not sufficiently clear in the current DREAM design it is assumed that the incident particle energy does not exceed the threshold energy of SiC at the divertor plate, and does not exceed the threshold energy of W at the baffle plate.

2.4. ARIES-AT

The ARIES-AT power core (shown in Fig. 10) has been developed with the overall objective of achieving high performance while maintaining attractive safety features, simple design geometry, credible maintenance and fabrication processes, and reasonable design margins as indications of reliability [7,24]. The design is based on Pb–17Li as breeder and coolant and a SiCf/SiC composite as structural material. The Pb–17Li first flows through the divertor, which is designed to accommodate a 5 MW/m² peak heat flux, and then through the blanket. To minimize waste and to decrease the cost, the blanket is subdivided radially into two regions: a replaceable first zone (in the inboard and outboard) and a life-of-plant second zone (in the outboard).

2.4.1. Power cycle

The Brayton cycle offers the best near-term possibility of power conversion with high efficiency. It is chosen to maximize the potential gain from high temperature operation of the Pb–17Li, which after exiting the blanket is routed through a heat exchanger with He as secondary fluid [25]. The Brayton cycle considered is shown in Fig. 11. It includes three-stage compression with two intercoolers and a high efficiency recuperator. Its main parameters are set under the assumption of state of the art components and/or with modest and reasonable extrapolation and are as follows:

- lowest He temperature in the cycle (heat sink) = 35°C
- turbine efficiency = 93%
- compressor efficiency = 90%
- recuperator effectiveness = 96%
- He fractional pressure drop in out-of-vessel cycle = 0.025

The maximum He cycle temperature is 1050°C, resulting in a high cycle efficiency of about 58.5%.

2.4.2. Blanket

As illustrated in Figs. 12 and 13, the blanket design is modular and consists of a simple annular box through which the Pb–17Li flows in two poloidal passes. The first pass is a high-velocity flow through the annular channel region keeping the box walls cooled. The coolant then turns and flows very slowly as a second pass through the large inner channel from which the Pb–17Li exits at a high temperature. This flow scheme enables operating Pb–17Li at a high outlet temperature (1100°C) while maintaining the SiCf/SiC composite and the SiC/PbLi interface at a lower temperature (about 1000°C).

The SiCf/SiC parameters and properties used in the analysis are consistent with the suggestion from the January 2000 International Town Meeting on SiCf/SiC Design and Material Issues for Fusion Systems [26] and are summarized in Table 5. Even though the SiCf/SiC provides for insulating the walls, thereby minimizing MHD effects,
the thermal analysis conservatively assumes MHD-laminar flow of Pb–17Li in the blanket and heat transfer by conduction only.

The temperature profile through the blanket has been estimated by a 2D moving coordinate analysis following the Pb–17Li flow through the first-pass annular wall channel and then through the second-pass large inner channel. The annular wall rib spacing is used as MHD flow control to promote a higher flow rate through the first wall (with larger toroidal spacing) than through the side and back wall. For example, for the first outboard blanket region shown in Fig. 12, having three channels in the first wall and 13 in the back wall allows for a high velocity of 4.2 m/s in the first wall channels and a lower velocity of 0.66 m/s in the back wall channel for the same MHD pressure drop. The second poloidal pass of the Pb–17Li through the large inner channel is much slower, with an average velocity of 0.11 m/s.

Fig. 14 illustrates the results for the parameters shown in Table 6. The maximum CVD-SiC temperature at the first wall (radial distance = 0) is 1009°C, the maximum SiC/Pl temperature is 996°C, and the maximum SiC/Pb–17Li interface temperature at the inner channel wall is 994°C. In this regard, and as discussed in Section 3.3, the chemical compatibility of Pb–17Li and SiC at high temperature is an area in need of R&D since only one experimental datum exists in the open literature for SiC (of unspecified purity) exposed to static Pb–17Li at 800°C for over 1500 h.

In addition to the illustrative thermal analysis described above, detailed structural, neutronics and safety design analyses of the blanket have been performed and can be found in Refs. [7,24,27–29]. Some of the main design features and performance parameters are highlighted below.

1. A 3D neutronics analysis yields a tritium breeding ratio of 1.1.
2. The 0.5-cm module side walls are pressure balanced except for the two walls at each end of a segment. These walls are thicker (about 2 cm) to accommodate the Pb–17Li pressure of 1 MPa.
3. The maximum SiC/Pl pressure + thermal stress is well within 190 MPa in the first wall as well as in the separation wall.
4. The Pb–17Li pressure drop through the blanket is about 0.24 MPa.
5. Safety analyses show that the low decay heat of SiC enables accommodation of any LOCA or LOFA scenarios without serious consequences to the blanket structure.
6. Maintenance methods have been investigated that allow for end-of-life replacement of individual components.
7. The 0.5-cm module side walls are pressure balanced except for the two walls at each end of a segment. These walls are thicker (about 2 cm) to accommodate the Pb–17Li pressure of 1 MPa.
8. An annular Pb–17Li coolant piping is used to service the blanket, with the lower temperature inlet flow in the annular channel and the higher temperature outlet flow in the inner channel. In this way any effect of the high SiC/Pb–17Li interface temperature on the inner channel wall would only result in a leak to the annular channel, which is tolerable. However, the structural integrity of the configuration would be ensured by the low temperature outer channel.
9. The fabrication scheme proposed for the blanket only three radial/toroidal coolant-containment brazes per module, as illustrated by the following fabrication steps for an outboard segment consisting of six modules:
   - Manufacturing separate halves of the SiC/SiC poloidal module by SiC weaving and SiC CVI or polymer processes;
   - Sliding each half module over the free-floating inner separation wall;
   - Brazing the two half modules together at the mid-plane;
   - Brazing the module end cap at the upper poloidal end;
   - Forming a segment by brazing six modules together (this is a bond that is not in contact with the coolant); and
   - Brazing the annular manifold connections to the lower poloidal end of the segment.

2.4.3. Divertor

Pb–17Li has a relatively low thermal conductivity and tends to offer limited heat removal performances, in particular in the presence of a magnetic field. In order to accommodate MHD effects, the proposed design:
- minimizes the interaction parameter ($< 1$), which represents the ratio of MHD to inertial forces;

Fig. 12. Mid-plane cross section of ARIES-AT outboard blanket modules (radial dimension in m).
Table 5
Suggested SiCf/SiC parameters and properties for design analysis of SiCf/SiC-based power plants for the long term (20–30 years in the future, or more)*

<table>
<thead>
<tr>
<th>Key SiC/SiC properties and parameters</th>
<th>Suggested value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density</td>
<td>≈ 3000 kg/m³</td>
</tr>
<tr>
<td>Porosity</td>
<td>≈ 5%</td>
</tr>
<tr>
<td>Young’s modulus</td>
<td>≈ 200–300 GPa</td>
</tr>
<tr>
<td>Poisson’s ratio</td>
<td>0.16–0.18</td>
</tr>
<tr>
<td>Thermal expansion coefficient</td>
<td>4 × 10⁻⁶ K⁻¹</td>
</tr>
<tr>
<td>Thermal conductivity in plane</td>
<td>≈ 20 W/m-K</td>
</tr>
<tr>
<td>Thermal conductivity through thickness</td>
<td>≈ 20 W/m-K</td>
</tr>
<tr>
<td>Maximum allowable combined stress</td>
<td>≈ 190 MPa</td>
</tr>
<tr>
<td>Max. allowable temperature (swelling</td>
<td>≈ 1000°C</td>
</tr>
<tr>
<td>basis)</td>
<td></td>
</tr>
<tr>
<td>Maximum allowable SiC/LiPb interface</td>
<td>≈ 1000°C</td>
</tr>
<tr>
<td>temperature</td>
<td></td>
</tr>
<tr>
<td>Min. allowable temperature (thermal</td>
<td></td>
</tr>
<tr>
<td>conductivity basis)</td>
<td></td>
</tr>
<tr>
<td>Max. allowable SiC burnup or other</td>
<td>Design</td>
</tr>
<tr>
<td>lifetime parameter</td>
<td>dependent</td>
</tr>
<tr>
<td>Cost</td>
<td>≤ $400/kg</td>
</tr>
</tbody>
</table>

*From the January 2000 International Town Meeting on SiCf/SiC Design and Material Issues for Fusion Systems [26].

Fig. 13. Cross section of an outboard ARIES-AT blanket module (all dimensions in cm).

- directs the flow in the high heat flux region parallel to the toroidal magnetic field;
- minimizes the Pb–17Li flow path and residence time in the high heat flux region.

The design is illustrated in Fig. 15, which shows a cross section of a divertor plate. It consists of a number of 2 cm × 2.5 cm SiCf/SiC poloidal channels. The front SiCf/SiC wall is very thin (0.5 mm) in order to maintain the maximum temperature and combined stress limits to < 1000°C and < 190 MPa, respectively. A 3.5-mm plasma-facing layer of W is bonded to the thin SiCf/SiC to provide additional structure to accommodate the 1.8-MPa Pb–17Li pressure and to provide sacrificial armor (about 1 mm).

In each channel a T-shaped flow separator is inserted. The Pb–17Li flows poloidally through one half of the channel, which acts as an inlet header. The flow is then forced to the plasma-facing region through small holes at one side of the channel. The flow through these small holes is inertial with an interaction parameter < 1. The Pb–17Li then flows toroidally to cool the high heat flux region through a very short flow path (2 cm). It is then routed back to the other side of the

Fig. 14. 2D moving coordinate thermal analysis of Pb–17Li flow through an outboard blanket module.
Table 6
ARIES-AT outboard blanket parameters assumed for the thermal analysis

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Geometry from plasma side:</td>
<td></td>
</tr>
<tr>
<td>CVD SiC thickness</td>
<td>1 mm</td>
</tr>
<tr>
<td>SiCf/SiC thickness</td>
<td>4 mm</td>
</tr>
<tr>
<td>SiCf/SiC thermal conductivity</td>
<td>20 W/m-K</td>
</tr>
<tr>
<td>LiPb channel thickness</td>
<td>4 mm</td>
</tr>
<tr>
<td>SiCf/SiC separation wall thickness</td>
<td>5 mm</td>
</tr>
<tr>
<td>SiCf/SiC separation wall thermal conductivity</td>
<td>6 W/m-K</td>
</tr>
<tr>
<td>Pb–17Li parameters</td>
<td></td>
</tr>
<tr>
<td>Pb–17Li velocity in first wall channel</td>
<td>4.2 m/s</td>
</tr>
<tr>
<td>Pb–17Li velocity in inner channel</td>
<td>0.11 m/s</td>
</tr>
<tr>
<td>Pb–17Li inlet temperature</td>
<td>764°C</td>
</tr>
<tr>
<td>Poloidal neutron wall load and plasma</td>
<td>Refs. [27,28]</td>
</tr>
<tr>
<td>Heat load conditions</td>
<td></td>
</tr>
<tr>
<td>(assuming no radiation from divertor)</td>
<td></td>
</tr>
</tbody>
</table>

Poloidal channel serving as an outlet header. The Pb–17Li velocity through the toroidal plasma-facing channel can be adjusted by changing the dimension of this channel or by increasing the number of toroidal passes through a plate. The reference design uses a 2-mm channel and a 2-pass flow, resulting in a 0.35 m/s velocity in the toroidal channel and a 0.06 s residence time.

The 2D moving coordinate method was used for the flow analysis under the conservative assumption of MHD-laminarized flow in the toroidal channel. Fig. 16 shows the typical results for a case with an inlet Pb–17Li temperature of 653°C, a W thickness of 3 mm, a SiCf/SiC first wall thickness of 0.5 mm, a Pb–17Li toroidal channel thickness of 2 mm, a SiCf/SiC inner wall thickness of 0.5 mm, a Pb–17Li velocity of 0.35 m/s and a surface heat flux of 5 MW/m². The maximum W temperature is about 1150°C, and the maximum SiCf/SiC temperature is about 950°C.

The resulting Pb–17Li pressure drop through the divertor based on the proposed flow configu-
ration is estimated at about 0.7 MPa. This is significantly larger than the blanket pressure drop. To minimize pressure stresses in the piping and blanket system, the Pb–17Li in the inlet manifold is kept at 1.1 MPa. The Pb–17Li is then pressurized to 1.8 MPa just before flowing though the divertor by means of an E–M pump making use of the existing toroidal magnetic field. The outlet flow from the divertor then rejoins the blanket inlet manifold at about 1.1 MPa.

The divertor fabrication scheme also aims at minimizing brazing and can be summarized as follows:

1. manufacture separate SiCf/SiC toroidal halves of the divertor plate by SiCf weaving and SiC CVI or polymer processes; maintain constant channel toroidal dimensions but tapered side wall thicknesses to account for torus geometry;
2. insert the inner SiCf/SiC separation wall in each divertor channel;
3. braze the two toroidal halves of the divertor plate together;
4. braze the end cap and manifold on each end;
5. bond the W layer to the SiCf/SiC front wall by plasma spray.

3. Material status

The materials research program attempts to develop and improve candidate materials with prospects to fulfill the design requirements laid out in conceptual design studies. It provides property data both for the existing state-of-the-art materials as well as reasonable extrapolations for the time frame in which fusion energy is expected to enter the marketplace. The key material development tasks include: development of fabrication and joining techniques for high quality and performance SiCf/SiC; establishment of base properties; determination of irradiated properties; and determination of stress and temperature limits and material lifetime.

The current trend in SiCf/SiC composite development is to improve stoichiometry both with the fiber and the matrix. Chemical vapor infiltration produces a stoichiometric, crystalline β-SiC matrix material. A similar matrix can be produced with the polymer impregnation & pyrolysis (PIP) process but this requires that the fibers have the thermal stability to withstand the high-temperature pyrolysis process needed to convert the polymer to crystalline β-SiC. As matured products, the newer fibers such as Tyranno-SA®, Hi-Nicalon®, S and MER corporation fiber [26] are expected to have the needed thermal stability so that SiCf/SiC composites with these fibers and a PIP matrix is a possibility. Other processes include chemical vapor reaction (CVR), reaction sintering (RS) and combined processing. At this stage, it is not clear which is the most attractive based on performance and cost, but they should be prioritized, given the limited R&D resources.

The interface materials used to provide the desired bonding between the fibers and matrix include C, SiC and BN. C, SiC and C/SiC interfaces are being developed and evaluated for their radiation and chemical stability in fusion environments, while BN does not look attractive because of the helium generation from the B.

Although a 3D architecture is preferred for fusion applications, it is not the focus of current development and experimental work mostly due to the difficulty of infiltration in a conventional architecture to provide low porosity. However, SNECMA’s experience indicates that low porosity can be achieved (≤ 5%) by non-orthogonal 3D weaving for enhanced infiltration (by chemical vapor infiltration, CVI) [26]. If required for hermeticity reasons, the porosity can also be lowered by subsequent melt infiltration (having no structural function). The use of new 3D SiCf/SiC composites with a higher resistance to inter-laminar shear stresses is very attractive and is described in more detail in the next section.

A variety of methods can be applied for joining SiCf/SiC to SiCf/SiC, including reaction bonding/sintering, polymer-based methods, brazing, stitching and mechanical joints. However, the R&D is still in an early stage in this area and much remains to be done to develop reliable and practical joining techniques, as described in the next section.

Cost is a key criterion when considering SiCf/SiC in fusion applications. By analogy with the evolution of the price of carbon fiber composites,
which has decreased substantially over the last 10 to 15 years, a Japanese study indicates that about one order of magnitude reduction in cost could be achieved with large-scale production of SiC\textsubscript{f}/SiC [26]. This assumes a large market for SiC\textsubscript{f}/SiC and a production process benefiting from the economy of scale. In this sense, among currently available advanced fibers, Tyranno-SA would benefit much more from this economy of scale than Hi-Nicalon, for example. Based on current cost, it seems reasonable to use $400 per kg of SiC\textsubscript{f}/SiC for power plant studies, given a lead time of $\geq$ 20 years for development.

Since the cost of the fibers represents about 50\% of the total cost, a lower-cost manufacturing technique is to start with a cheaper carbon fiber and to transform it into a SiC\textsubscript{f}/SiC composite through CVR. This is being investigated by MER [26]. Such an approach could reduce the SiC\textsubscript{f}/SiC cost to about $100 per kg or less. A key issue is the completeness of the conversion process and the closeness of the final product to stoichiometry since this would greatly impact the performance and properties of the final product.

### 3.1. Fabrication and joining

It is difficult to manufacture a complicated blanket shape (curved parts, stiffeners) with traditional 2D SiC\textsubscript{f}/SiC processes due to delamination problems. As a consequence, the use of new 3D SiC\textsubscript{f}/SiC composites with a higher resistance to inter-laminar shear stresses seems inevitable. The Cerasep\textsuperscript{®} N3-1 produced by SNECMA by CVI uses Nicalon\textsuperscript{®} NL207 fibers. This composite is based on an advanced texture called GUIPEX\textsuperscript{®}, a multilayer 3D texture [9] in which a third texture direction, created by linking layers during weaving, minimizes delamination risks when shaping and densifying the parts. The main properties of Cerasep\textsuperscript{®} N3-1 are reported in Table 1. However, the maximum thickness of the 3D composite is around 6 mm for the present CVI state of the art techniques. The limits of this composite for fusion power core applications are mainly linked to its low thermal conductivity and its resistance to irradiation damage. The use of low oxygen content fibers like Hi-Nicalon fibers (Nippon Carbon) could help alleviate those two problems. Also, the maximum operating temperature based on strength degradation increases from about 1100°C to about 1300°C. A new composite, Cerasep\textsuperscript{®} N4-I, which uses Hi-Nicalon fibers is currently under development by SNECMA. This new composite appears very promising for fusion power core applications, but no data are available.

In order to use SiC\textsubscript{f}/SiC as a structural material, reliable and practical joining techniques are required. A variety of methods can be used for joining SiC\textsubscript{f}/SiC to SiC\textsubscript{f}/SiC. Although some of them have already been industrialized, methods that may qualify for fusion blanket assembly are still in a developmental stage. Current efforts focus mainly on establishing a technical basis for joints with improved and tailorable properties rather than on demonstrating specific properties for specific joining techniques. The effect of neutron irradiation, which is one of the most important technical issues for application to fusion blankets has still not been evaluated. Methods of joining SiC\textsubscript{f}/SiC to SiC\textsubscript{f}/SiC can be classified into SiC-based, non-SiC-based and mechanical joints. The first category includes reaction bonding and pre-ceramic polymer adhesives. The advantages of SiC-based joints are their coefficients of thermal expansion (CTE), generally comparable to SiC\textsubscript{f}/SiC, excellent thermo-chemical stability and the straightforward applicability of techniques developed for monolithic SiC production and matrix densification of SiC\textsubscript{f}/SiC. The second category includes glass–ceramic joints and metal-based brazes. These methods are generally easy to apply and to disassemble since they do not incorporate the production of refractory SiC, though their reactivity with SiC\textsubscript{f}/SiC and the consequences (CTE mismatch effects and thermo-chemical stability) need to carefully be examined. Examples of the last category are mechanical fastening and fabric co-stitching. These methods may be used in combination with any adhesive-type joint. Techniques of fairly complex shaping of fiber fabrics have been developed in recent advanced motor generator programs in order to form near-net-shape gas turbine components, and the techniques like co-stitching might be classified into the ‘fabric shaping’ category.
3.1.1. Reaction bonding

Presently, reaction bonding is a most promising method for permanently joining SiC/SiC. In this process, molten silicon reacts with carbon in fine powder or porous skeleton shapes to produce SiC. Reaction bonding appears advantageous because it produces fully crystalline SiC with a very high strength, high thermal conductivity and excellent creep properties, in addition to other inherently attractive properties of SiC. Flexural strength and thermal conductivity of approximately 1 GPa and 130 W/m K, respectively, have been demonstrated for a monolithic SiC developed by the CREST-ACE project [30] through a process that incorporates a molten silicon infiltration to C–SiC powder complex. Several other reaction bonding techniques have been successfully developed, including a robust technique called ARCJoinT [31]. A potential disadvantage of the reaction bonding is that it requires a heat treatment typically at about 1700 K, or a temperature over the melting point of silicon, which is harmful to the older generations of SiC-based fibers. However, this may not be a problem for SiCf/SiC using advanced fibers that survive up to over 2000 K [32].

3.1.2. Pre-ceramic polymers

A pyrolyzed product of pre-ceramic polymers is generally a mixture of nano-crystalline SiC, amorphous silica and amorphous or glassy carbon with a substantial porosity. The mostly-amorphous Si–C–O material has a reduced strength, a very low thermal conductivity and inferior thermo-chemical and radiation stabilities. However, the properties of polymer-derived SiC are rapidly improving with recent extensive efforts. A near-stoichiometry composition will be achieved by using polymer precursors that are inert and designed for stoichiometric composition. Polymer precursors for improved stoichiometry, e.g. polymethylsilane by Ube industries and AHPCS by Starfire Systems Corporation, are becoming commercially available [33]. A precursor for both stoichiometry and reduced oxidization susceptibility is being developed in the CREST-ACE program. The mechanical properties of SiCf/SiC joints with AHPCS and other pre-ceramic polymers are under evaluation in a collaborative work between Kyoto University and Pacific Northwest National Laboratory (PNNL).

3.1.3. Glass–ceramic joints

Glass–ceramic is a relatively new material that consists of glass and ceramic. The benefits of this material include tailorable properties, oxidation resistance and mechanical strength provided by the ceramic phases [34]. As one of the options for SiCf/SiC joints for fusion, Politecnico di Torino and Kyoto University are jointly studying a calcia–alumina (CA) glass ceramic. They have so far demonstrated (in their first trial) an excellent wettability of CA with crystalline SiC and shear strength of about 30 MPa for a pressureless joint [35]. The glass ceramic joints will allow quick and non-permanent use, since they may easily be decomposed at temperatures exceeding the softening temperature of the glassy phase. Applicability to permanent joints relies on future development, including the evaluation of neutron irradiation effects.

3.1.4. Brazing

A promising brazing technique has been recently proposed using a brazing process already dedicated for refractory materials to SiCf/SiC. The braze material, Brasic®, is compatible with SiC and is being developed by CEA, France [36]. The major advantages of the filler alloys, composed of low-activation elements, are: high temperature resistance (1400°C), no chemical reaction with SiC, good wetting and good adherence to the interface (mechanical shear strength as high as 80–100 MPa up to 800°C). Brasic® alloys ensure at the same time good mechanical bonding and a perfect tightness of the brazed junction. By using different alloys and compositions (Brasic H2 and V2 or V3) and brazing in vacuum or in inert atmosphere it has been possible to control the infiltration of the alloy. Use of Brasic V3 for joining at 1300°C in a neutral atmosphere resulted in a sound joint with perfect filling of the joint gap and without infiltration in the composite, as shown in Fig. 17. For this joint, a shear strength of 174 MPa was obtained at room temperature and about 100 MPa at 800°C. The main limitation of the method results from the free Si content
and the open high porosity of the composite. This low-activation brazing alloy can in principle be used also for coating SiC/\text{SiC} composites.

3.2. Properties

3.2.1. Fiber properties

Fibers are the backbone of the continuous fiber ceramic matrix (SiC/\text{SiC}) composites being considered for fusion energy applications [37–39]. Even though composite properties are controlled by several factors besides fiber strength, such as volume fraction of fibers, fiber/matrix interface structure, fiber weave architecture and matrix properties, there are key fiber properties that control radiation resistance, high-temperature strength and stability and weaveability. A summary of the SiC fibers being considered for fusion energy applications is given in Table 7.

The fiber diameter and elastic modulus determine the weaveability of the fiber; consequently, smaller diameter fibers are needed with a high elastic modulus. Coarser fiber weave architecture must be used for high elastic modulus fibers. The Nicalon and Tyranno fibers are derived from polymers with a trend towards higher fiber purity, crystallinity, stoichiometry and elastic modulus in the development of Nicalon fibers. Dow-Corning no longer makes a SiC fiber, but this fiber is listed as typical of those made from sintered \textbeta-SiC crystals. The MER fiber is made by chemical conversion of C fibers with the SiC component of the fiber being crystalline, stoichiometric, \textbeta-SiC; however, these fibers can be made with a varying degree of conversion so they may have a C core and a SiC outer layer.

The creep strength of SiC fibers is directly related to the density, degree of crystallinity and stoichiometry. The bend stress relaxation test is a simple test for comparing the creep properties of fibers as shown in Fig. 18. This figure shows a relaxation ratio \( m \) as a function of inverse temperature, where no creep corresponds to an \( m = 1 \) and increasing creep strain corresponds to decreasing values of \( m \). There is a clear shift in the curves to higher temperatures, i.e. lower reciprocal temperatures with increasing fiber density, crystallinity, stoichiometry and elastic modulus. The ranking in order of increasing creep strength is: Lox-M Tyranno, Nicalon 201, UF-Dev, Hi-Nicalon and Dow-Corning (Dev and NASA).

3.2.2. Composite properties

A summary of the properties of several commercially available SiC/\text{SiC} composites is given in Table 8. It should be stated that laboratory-scale composites with rapidly improving properties are currently being developed in a very short time cycle. From the table, the specific gravity ranges from a low of 2100 to a high of 2700 kg/m\(^3\) while

Fig. 18. Bend stress relaxation (BSR) test of SiC-based fibers in Argon [40,41].

Fig. 17. Brasic\textsuperscript{®} joint micrography.
Table 7
Properties of as-produced silicon-based fibers

<table>
<thead>
<tr>
<th>Trade name</th>
<th>Composition and structure</th>
<th>Dia. (μm)</th>
<th>Avg. strength (GPa)</th>
<th>Modulus (GPa)</th>
<th>Density (g/cm³)</th>
<th>Thermal expansion coeff. b</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nicalon®</td>
<td>65% SiC, 23% SiO₂, 11% C</td>
<td>14</td>
<td>3.0</td>
<td>220</td>
<td>2.55</td>
<td></td>
</tr>
<tr>
<td></td>
<td>(~ 3 nm β-SiC grains)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Hi-Nicalon®</td>
<td>77% SiC, 23% C, &lt;0.5% O</td>
<td>14</td>
<td>2.8</td>
<td>270</td>
<td>2.74</td>
<td></td>
</tr>
<tr>
<td></td>
<td>(10–100 nm β-SiC grains)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Nicalon-S</td>
<td>95% SiC, 5% C (10–100 nm β-SiC grains)</td>
<td>14</td>
<td>2.6</td>
<td>420</td>
<td>3.1</td>
<td></td>
</tr>
<tr>
<td>Dow-Sylramic</td>
<td>&gt;99% SiC (30 nm β-SiC grains)</td>
<td>10</td>
<td>2.6</td>
<td>420</td>
<td>3.1</td>
<td>4.6</td>
</tr>
<tr>
<td>Tyranno-SA</td>
<td>&lt;0.5%Al, balance SiC</td>
<td>8</td>
<td>2.8</td>
<td>400</td>
<td>3.1</td>
<td>4.6</td>
</tr>
<tr>
<td></td>
<td>(50–500nm β-SiC grains)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>[MER]</td>
<td>β-SiC</td>
<td>5</td>
<td>–</td>
<td>340–410</td>
<td>2.9–3.1</td>
<td>4.6</td>
</tr>
</tbody>
</table>

a Although Nicalon and Hi-Nicalon are not considered for fusion applications, they are included as standard fibers.
b Average from 20 to 100°C.
Table 8
Properties of several SiC/SiC composites

<table>
<thead>
<tr>
<th>Property</th>
<th>Units</th>
<th>SNECMA (Cerasep® N2-1)</th>
<th>DuPont</th>
<th>DuPont enhanced SiC/SiC</th>
<th>DOW/Kaiser S-201</th>
<th>DuPont Hi-Nicalon</th>
<th>DuPont Nicalon-S</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>296 K</td>
<td>1273 K</td>
<td>296 K</td>
<td>1273 K</td>
<td>296 K</td>
<td>1273 K</td>
</tr>
<tr>
<td>Fiber content</td>
<td>%</td>
<td>40</td>
<td>40</td>
<td>40</td>
<td>40</td>
<td>40</td>
<td>40</td>
</tr>
<tr>
<td>Specific gravity</td>
<td>kg/m³</td>
<td>2500</td>
<td>2500</td>
<td>2500</td>
<td>2500</td>
<td>2300</td>
<td>2300</td>
</tr>
<tr>
<td>Porosity</td>
<td>%</td>
<td>10</td>
<td>10</td>
<td>10</td>
<td>10</td>
<td>4</td>
<td>4</td>
</tr>
<tr>
<td>Tensile strength</td>
<td>MPa</td>
<td>285</td>
<td>285</td>
<td>200</td>
<td>228</td>
<td>228</td>
<td>254</td>
</tr>
<tr>
<td>Elongation (tensile)</td>
<td>%</td>
<td>0.65</td>
<td>0.75</td>
<td>0.22</td>
<td>0.28</td>
<td>0.41</td>
<td>0.61</td>
</tr>
<tr>
<td>Young’s modulus (tensile)</td>
<td>GPa</td>
<td>230</td>
<td>200</td>
<td>215</td>
<td>226</td>
<td>141</td>
<td>145</td>
</tr>
<tr>
<td>Flexural strength</td>
<td>MPa</td>
<td>300</td>
<td>400</td>
<td>300</td>
<td>400</td>
<td>345</td>
<td>–</td>
</tr>
<tr>
<td>Compression strength</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>In plane</td>
<td>MPa</td>
<td>580</td>
<td>400</td>
<td>580</td>
<td>480</td>
<td>503</td>
<td>–</td>
</tr>
<tr>
<td>Through the thickness</td>
<td>MPa</td>
<td>420</td>
<td>380</td>
<td>420</td>
<td>380</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>Shear strength</td>
<td>MPa</td>
<td>40</td>
<td>35</td>
<td>40</td>
<td>35</td>
<td>31</td>
<td>–</td>
</tr>
<tr>
<td>Thermal diffusivity</td>
<td></td>
<td>× 10⁻⁶ m²/s</td>
<td>12</td>
<td>5</td>
<td>12</td>
<td>5</td>
<td>–</td>
</tr>
<tr>
<td>In plane</td>
<td></td>
<td>× 10⁻⁶ m²/s</td>
<td>6</td>
<td>2</td>
<td>6</td>
<td>2</td>
<td>1.5^a</td>
</tr>
<tr>
<td>Through the thickness</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Coefficient of thermal expansion</td>
<td></td>
<td>× 10⁻⁶ K⁻¹</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>2.9</td>
</tr>
<tr>
<td>In plane</td>
<td></td>
<td>× 10⁻⁶ K⁻¹</td>
<td>2.5</td>
<td>2.5</td>
<td>2.5</td>
<td>2.5</td>
<td>2.9</td>
</tr>
<tr>
<td>Through the thickness</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Fracture toughness</td>
<td>MPa m¹/²</td>
<td>30</td>
<td>30</td>
<td>30</td>
<td>30</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>Specific heat</td>
<td>J/kg K</td>
<td>620</td>
<td>1200</td>
<td>620</td>
<td>1200</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>Total</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>emissivity</td>
<td></td>
<td>0.8</td>
<td>0.8</td>
<td>0.8</td>
<td>0.8</td>
<td>0.8</td>
<td>0.8</td>
</tr>
</tbody>
</table>

^a Thermal conductivity (W/m K).
the porosity ranged from a low of 4% to a high of 10%. The elastic moduli closely reflect the modulus of the fibers. This is particularly noticeable for the DuPont composite with Hi-Nicalon fibers. The composite elastic modulus is 270 GPa and the Hi-Nicalon fiber modulus is also 270 GPa. The crystalline β-SiC matrix produced by CVI has a modulus approaching that of monolithic SiC except for the effect of porosity. The sum of these factors has resulted fortuitously in a composite with an elastic modulus equal to the fiber modulus. The tensile strengths of these materials are less than the flexural strengths because of the greater compressive strengths of these materials. The elastic modulus of the DuPont composite material is decreased substantially from the CVI-SiCf/SiC composite material because of the additive used to improve the stress–rupture properties in high-oxygen concentration environments. The strength of these materials is relatively insensitive to temperature up to the limits of fiber stability. Materials made with the newer, more thermally stable fibers will retain their strengths to higher temperatures than those made with the older, less thermally stable fibers. Some increase in flexural strength occurs at the composite fabrication temperature relative to ambient temperature. Fracture toughness values for SiCf/SiC composites are also listed in Table 8, showing a value of 30 MPa m^{1/2}. This is a $K_Q$ value (a measure of the ultimate failure load of a notched specimen) and not a $K_{IC}$ value (plane-strain critical stress intensity) because these materials do not fail by propagation of an unstable crack; rather, they exhibit $R$ curve or rising load failure. The toughness value is highly dependent on the fiber/matrix interface properties and weakly dependent on the fiber and matrix properties. As with the tensile and flexural strength, the fracture toughness is temperature independent up to the fiber instability temperature.

Through thickness thermal diffusivity values (as shown in Table 8) range from about $12 \times 10^{-6}$ m$^2$/s at room temperature to about $2 \times 10^{-6}$ m$^2$/s at 1000°C (corresponding to thermal conductivity values of about 20 and about 6 W/m K, respectively) for material made by CVI. Much smaller values are given for the DOW/Kaiser S-201 material made by the PIP process, which results in an amorphous matrix with low thermal conductivity. The highest room temperature thermal conductivity has been reported for a MER CVR processed composite with a value of 41 W/m K as compared with a value of 6 W/m K for CVI processed material. The MER material is processed from C/C composites and the fibers retain some fraction of graphite, which increases the thermal conductivity.

Both the composite creep rate and the time-dependent crack growth rate are dependent on the fiber creep rates [42–44]. DiCarlo and Yun [42] showed a decreasing creep rate for composite material with Nicalon, Hi-Nicalon and Sylramic fibers, respectively. At 1400°C, there was a decrease in the creep rate by a factor of 2.5 for these three fibers. Henager and Jones [43,44] likewise demonstrated that the crack growth rate of composites is directly dependent on the fiber creep rate and that the crack growth rate of composite material made with Nicalon fibers is equal to that of composite material made with Hi-Nicalon fibers but tested at a 75°C higher temperature. This temperature shift is identical to that shown in Fig. 18 for these two fibers.

Thermal fatigue and thermal shock are important properties of ceramic composites for fusion energy systems: thermal fatigue because of heat-up/cool-down cycles, and thermal shock because of plasma discharge and rapid heating of first wall materials. Clearly, the magnitude of the thermal spike in blanket materials decreases with increasing distance from the first wall. The fatigue and thermal shock behaviors of ceramics have been reviewed by Jones et al. [37]. The thermal fatigue of a Nicalon-fiber-reinforced SiCf/SiC composite was evaluated by Jones and Henager [45] using low-cycle mechanical fatigue. They found that the crack velocity decreases with increasing fatigue cycles in a manner similar to that observed for sub-critical crack growth with time. There was a factor of 10 decrease in the load cycle between the first and 25th cycles at a stress intensity of 18 MPa m^{1/2}. The sub-critical crack growth decreases with time because additional bridging (i.e. load carrying) fibers accumulated behind the crack tip. At higher stress intensities the crack velocities
shown that, at 1200°C, above this threshold stress, Holmes [46] has shown that the matrix cracking stress is approximately equal to the matrix cracking stress. This stress threshold is exhibited below which crack growth does not occur. This stress threshold is below which crack growth does not occur. This stress threshold is approximately equal to the matrix cracking stress.

Above this threshold stress, Holmes [46] has shown that, at 1200°C and 10 Hz, the number of cycles to failure is a function of the $R$ ratio and peak stress. For the material studied by Holmes [46], the matrix cracking stress was 200 MPa and the ultimate strength was 380 MPa at 1200°C.

The thermal shock response of materials is very dependent on the temperature change and the change rate and the constraint on the sample or component. For continuous fiber–ceramic–matrix composites, the thermal shock behavior also depends on whether the temperature change is positive or negative. Most thermal shock data are for samples quenched from an elevated temperature; however, in a fusion reactor a plasma discharge will result in rapid heating. Ceramic composite properties are not strain rate dependent, other than in the creep strain rate range, so the heating rate should not be a concern. Thermal shock tests with a positive temperature change were performed by Eckel et al. [49] and Wang and Singh [50]. Eckel measured the thermal shock of monolithic and composite material heated from 1300 to 2300°C in a burner rig. The monolithic material failed after 1.5 cycles while the SiC$_f$/SiC composite survived 25 cycles without a loss of strength. However, with a heating rate of 1900°C/s the composite exhibited a 35% strength loss as compared with no strength loss at a heating rate of 1700°C/s.

### 3.2.3. Properties of irradiated materials

Neutron irradiation of SiC induces changes in properties that are critical to its application in fusion energy systems. Displacement of Si and C atoms occurs with different displacement energies with C having the lower displacement energy and therefore the greater displacement rate. These atomic displacements result in the formation of dislocation loops parallel to the {111} crystallographic planes. These loops induce an isotropic dimensional change that saturates at neutron fluences of $2 \times 10^{20}$ n/cm$^2$ at temperatures below about 1000°C. The saturation swelling value becomes dependent only on the irradiation temperature below 1000°C once the saturation fluence has been achieved. A significant amount of data has been generated on the swelling of SiC for the high-temperature gas-cooled reactor, temperature monitors in irradiation experiments and now for the fusion materials program. A summary of this database is shown in Fig. 19, along with several curves fitted to the data. It is apparent that the swelling decreases with increasing temperature to a relatively small volume change of a few tenths of a percent at 1000°C. A density decrease accompanies the swelling response. Irradiation of SiC at temperatures above 1000°C results in tetrahedral voids bounded by {111} planes and the volume change is fluence-dependent, unlike irradiation below 1000°C [52]. The quantity and quality of the swelling data for SiC at temperatures above 1000°C is lower than that for temperatures below 1000°C; there is a need to better define the minimum temperature for void swelling, the fluence dependence and the magnitude of swelling. Dimensional changes in composite materials made by CVI and irradiated at temperatures below 1000°C are consistent with those for monolithic material [53] because the matrix defines the primary composite response. The CVI matrix is crystalline, stoichiometric β-SiC, so it follows the swelling curve in Fig. 19. Crystalline, stoichiometric β-SiC fibers such as the DOW Sylramic and MER fibers exhibit the same swelling response as shown in Fig. 20 (DOW-X fibers).

Polymer derived fibers and matrix materials (PIP) that have not been fully crystallized by thermal processing shrink during irradiation are shown in Fig. 20. While the database shown in the figure is for a range of temperatures it provides a clear representation of the response of crystalline and non-crystalline SiC fibers. Fiber shrinkage leads to fiber/matrix debonding as reported by Hollenberg et al. [53] and a decrease in elastic modulus and fracture strength. Therefore, there is a critical need to optimize the microstructure of SiC/SiC composites (i.e. fiber, fiber/matrix interphase and matrix) to minimize the interfacial...
stress between the fiber and matrix. The results in Fig. 20 show that the density of the crystalline and polymer-derived fibers approach each other with increasing neutron fluence.

Generation of point defects and dislocation loops in monolithic SiC causes phonon scattering and a commensurate decrease in the thermal conductivity. A summary of the ratio of the irradiated to unirradiated thermal conductivity is given in Fig. 21 where it can be seen that this ratio is about 0.5 for an irradiation temperature of 1000°C and decreases to 0.1–0.2 for an irradiation temperature of 400°C. This curve is the inverse of the swelling curve because defects
Irradiation effects on the flexural strength of SiCf/SiC composites for Nicalon-CG fiber reinforced material are shown in Fig. 22 up to a fluence of 80 dpa. This plot shows the four-point bend strength as a function of temperature and fluence for the available data along with some extrapolation to high temperatures. The maximum strength in both the unirradiated and irradiated conditions occurs at 800–1000°C with a drop of the bend strength from about 400 to about 200 MPa for the irradiated material at 800°C. Unirradiated four-point bend strengths for material with Hi-Nicalon fibers is about 500 MPa. The decrease in the fracture strength with irradiation has been coupled with the shrinkage of the polymer-derived, non-crystalline Nicalon-CG fiber (Fig. 20) and the subsequent debonding of the fiber from the matrix such that load-transfer to the fibers does not occur. Even with this debonding the four-point bend strength of a composite material irradiated to 80 dpa at 800°C is 200 MPa. The tensile strength will be about 66% of the flexural strength so the irradiated tensile strength is likely to be about 130 MPa. Snead [58] has measured the four-point bend strength of composite material reinforced with Hi-Nicalon and irradiated to 1 dpa (Fig. 23). These results clearly show the improved radiation resistance obtained with com-

![Graph](image_url)

**Fig. 20.** Density as a function of dose or fluence for four types of uncoated SiC-based fibers [54].

produced by irradiation results in phonon dispersion. Both monolithic and composite materials are included in this figure even though the fiber and fiber/matrix interface affect the composite thermal conductivity in addition to radiation-induced defects [56]. The Hollenberg and Senor composites are made with Nicalon-CG fibers, which have a very low thermal conductivity and therefore contribute little to the composite thermal conductivity.

![Graph](image_url)

**Fig. 21.** Thermal conductivity degradation as a function of irradiation and test temperature for crystalline SiC and SiCf/SiC composites with CVI-SiC matrix [55].
posites made with Hi-Nicalon and Nicalon-S fibers. The strength of irradiated material with Nicalon-S is equal to or slightly greater than the strength of unirradiated material.

Irradiation was shown by Price [59] and Scholz [60] to enhance the creep rate of β-SiC produced by CVD for SCS-6 fibers (Fig. 24). SCS-6 fibers are formed by CVD on a C core. It is clear that irradiation enhances the creep rate of crystalline β-SiC above that for thermal creep at temperatures ranging from 450 to >1000°C. The irradiation creep rate is independent of temperature over this range in a manner similar to that exhibited by metals. Little is known about the mechanisms involved in the irradiation enhanced creep of SiC. Also, the effect of fiber creep on composite creep can be described as shown earlier for crack growth controlled by thermal creep of fibers. However, for a CVI composite with a crystalline β-SiC matrix reinforced with Nicalon-S or other highly crystalline fibers the creep rate of the fiber and matrix will become equal. Under these circumstances, the composite may creep more like monolithic SiC as shown by Price, although a more thorough analysis that includes load transfer to the fibers and the stress dependence of creep of the matrix and fiber at different stresses must be completed before the creep behavior of SiCf/SiC composites can be fully described. In the extreme case where the fibers creep at the same rate as the matrix, the data (Fig. 24) of Price [59] can be used to estimate the irradiation-enhanced creep rate. This creep rate is $10^{-10} \text{ s}^{-1}$ at a stress of 200
plot (Fig. 25) of the transmutation rate for an ARIES first wall design [1] shows a very large build-up of He and H and a significant increase in the Mg concentration. The effect of these chemical changes on the properties of SiC and SiC/fiber composites is not known although some studies using He ion implantation followed by neutron irradiation or ion irradiation have been completed or are in progress. Helium release studies by Sasaki et al. [62] have demonstrated that He is mobile in the SiC lattice at temperatures as low as 250°C but that He trapped at crystal interfaces is immobilized up to about 1000°C. Significant helium release from crystal interfaces occurs at temperatures exceeding 1800°C. Therefore, it is likely that the helium generated by nuclear transmutation will be retained in the structure at crystal interfaces in the fiber, fiber/matrix interface or the CVI growth interfaces unless the He can escape through open pores.

3.2.4. Helium effects

Transmuted He gas effects have been simulated using He implantation or simultaneous He-ion and heavy-ion irradiation. Calculation of the total amount of transmuted He based on fusion reactor material design was carried out by Noda et al. [63]. The He/dpa ratio is predicted to be about 100 appm/dpa in the first wall region of the solid breeder blanket system when using SiC fiber composites as the structural material. Helium is an insoluble element in almost all the solid materials, and it stabilizes vacancy type clusters. It may enhance void swelling of SiC under fusion reactor conditions.

Accelerator irradiation experiments have been used to study helium effects. Single, dual, or triple-beam experiments have been conducted to investigate helium effects on microstructural evolution. The microstructural development of a composite containing advanced fiber fabrics was not clearly observed below 800°C, probably because the irradiation temperature was too low for helium to move and to cluster [64,65], but small clusters of defects were observed. Helium diffusion behavior and microstructural observation of SiC fiber composites, SiC fiber and monolithic SiC was reported by Hasegawa et al. [66]. Helium in

Fig. 24. The temperature dependence of the irradiation creep compliance determined for light ion, low dose (<0.07 dpa) and neutron, high dose (~7.7 dpa) irradiations of SCS-6 SiC fiber and CVD β-SiC strips (from Price), respectively [61].

Fig. 25. Silicon carbide transmutation products–ARIES [1] first wall irradiation at 4.7 MW/m².
carbon phases becomes mobile between 500 and 800°C. Helium in the SiC phase becomes mobile above 900 or 1000°C. In the higher temperature region, the He desorption rate from the SiC fibers becomes much larger than from monolithic SiC. This result is consistent with a previous diffusion study by Jung [67]. These results suggest that helium might be mobile in C and the amorphous like Si-C phase in the composite in the expected range of reactor operating temperature.

Mechanical property changes were investigated using high energy He implantation by cyclotrons [60,61]. Bending tests of SiCf/SiC composites after high temperature implantation (950°C) up to a helium concentration of 2500 appm showed a 38% mean decrease of strength. The load–displacement curve exhibited a significant decrease of the values of the maximum deflection [68]. Swelling of the SiC matrix and shrinkage of Nicalon fiber were reported by Scholz et al. [70]. A decrease in strength was observed at about 150 appm implantation at temperatures ranging from 400 to 800°C [69]. Microstructural observations of He-implanted SiCf/SiC composites were reported by Hasegawa et al., as shown in Fig. 26 [71]. Helium bubbles were not observed at the interface between the matrix and interphase(C), and between the fiber and interphase(C) or in the SiC fiber after about 10000 appm Helium implantation at room temperature and annealing at 1400°C. Bubbles were observed only at the grain boundaries in SiC. Helium desorption results can explain this microstructure.

3.2.5. Compatibility

3.2.5.1. SiC/PbLi. Only limited data on the compatibility of SiCf/SiC with Pb–17Li are available. Fenici and Scholz [72] reported the results of exposure of 2D CVI-SiCf/SiC composites to static Pb–17Li at 800°C for 1500 h. Except for a penetration of Pb–17Li into the composite through the open pores, the results indicated that the SiCf/SiC composites are practically inert in Pb–17Li. Tests, for relevant Pb–17Li velocity (up to 1 m/s), duration (10000 h), and high temperatures (about 1100°C) have yet to be performed.

3.2.5.2. SiC/ceramic breeders and SiC/Be. The compatibility of SiCf/SiC commercial composite with solid breeders (Li4SiO4 and Li2TiO3) under fusion relevant conditions was investigated by LaBarbera et al. [73]. The ceramic breeder Li4SiO4 was industrially produced by Glaswerke Schott Co. (Germany) and has an average diameter of 0.55 mm [74]; the Li2TiO3 pebbles have a shape and dimensions similar to the Li4SiO4 ones and have been manufactured by ENEA-Casaccia (Italy). The ceramic composite investigated was the 3D Cerasep® N3-1 (Nicalon CG + CVI SiC matrix) [75]; the samples were subjected to a final
uniform CVD SiC coating (about 100 µm thick). Three cells were operated with Li$_4$SiO$_4$ and three others with Li$_2$TiO$_3$ for 216, 1000 and 10000 h at 800°C in flowing He containing 1000 ppm H$_2$; downstream operating samples (so-called ‘blank’) and immersed samples were provided for each cell.

After 10000 h for the Li$_4$SiO$_4$ test, lithium metasilicate (Li$_2$SiO$_3$) and other complex silicates were found on blank samples indicating a lithium migration out of the reaction chamber. On immersed samples the original SiC coating was still present after exposure: its thickness ranged from 80 to 100 µm. A surface layer of about 30 µm of Li$_2$SiO$_3$ that was not adherent but heavily cracked was observed (Fig. 27). The transformation of the SiC coating is not limited to a protective film since a linear law of weight increase was observed, but no changes have been detected inside the specimens. Slight changes of the mechanical properties (Young’s modulus and flexural strength) were observed at different temperatures. Therefore, for Li$_4$SiO$_4$ exposure, a long lifetime can be foreseen for this type of composite depending on the SiC coating layer thickness.

On blank samples over Li$_2$TiO$_3$ only silica was found for <10000 h, while on immersed samples the original SiC coating was not present over about half of the surface and a heavily cracked layer of 80–100 µm, identified as various silicates and silica, was observed. The original SiC layer is full of voids and cracks across its entire thickness (Fig. 28). After 10000 h, SiC fibers are not protected any more and are exposed to the reacting environment. In spite of these observations mechanical properties appear not to be significantly affected, indicating that the bulk is not damaged after up to 10000 h. Therefore, since the SiC coating is consumed and the fibers are exposed to the gas phase, a rapid reduction of the mechanical properties is foreseen for longer exposure times.

The chemical reactivity of monolithic SiC with Li oxide breeder materials and beryllium was studied by Kleykamp by isothermal and anisothermal annealing [76]. Powder reaction experiments were made between α-SiC and Li$_4$SiO$_4$, LiZrO$_3$ and LiTiO$_3$ pellets in a quartz tube furnace at 700°C for two weeks under static Ar. No reaction between α-SiC and the Li ceramics breeders was observed after two weeks of isothermal annealing. Conversely, anisothermal annealing has shown that following exposure to:

- Li$_4$SiO$_4$: the α-SiC remained unaffected but at temperature higher than 1000°C the SiO$_2$ formed (following Li, O$_2$ and Li$_2$O evaporation from solid Li$_4$SiO$_4$) was found to react with Li$_4$SiO$_4$ to produce Li$_4$SiO$_3$;
- Li$_2$ZrO$_3$: a reaction between α-SiC and Li$_2$ZrO$_3$ with the production of ZrC and Li$_2$SiO$_3$ took place at temperatures >1100°C;
- Li$_2$TiO$_3$: a solid state reaction occurred between α-SiC and Li$_2$TiO$_3$ with the formation of TiC and Li$_2$SiO$_3$ at temperatures >1160°C.
The compatibility with Be was studied by annealing a β-SiC–Be pebbles mixture in a closed cylindrical refractory metal capsule between 700 and 900°C for ≤70 days. No indication of an incompatibility between Be and SiC was reported after the 700°C annealing, whereas, between 800°C and 900°C, a two-phase Be2C–Si was observed on the SiC pellet. The thickness of this phase increased with exposure time because Be2C and Si are in thermodynamic equilibrium at 1000°C. For this reason the development of a protective coating on SiC/SiC, able to prevent Be diffusion into the SiC at reactor operating conditions, will be required for breeding blankets with Be as multiplier.

3.3. Stress limits

3.3.1. Background

SiC/SiC composites presently envisaged for fusion reactors are still in an early phase of development. The ability to correctly assess their limits is therefore a key factor to provide guidelines for further R&D. The appropriate design methodologies have to account for the peculiarities of their mechanical behavior, in particular:

- Non-linearity of the stress–strain relation under tensile loading; this non-linearity is due to the progressive development of damage mechanisms inside the composite.
- Different mechanical properties depending on the loading direction. Even for 3D composites, the number of fibers through the thickness is lower and their arrangement different. Mechanical properties are generally worse along this direction.
- Different failure mechanisms and therefore different strengths under tensile and compressive loading. Under tensile loading failure is due to damage accumulation, while under compressive loading the mechanical behavior is brittle and elastic up to rupture.

The methodology commonly used is to consider two independent successive processes: deformation under load and failure. The stress field distribution is then obtained by means of specific behavioral models, and the resistance criteria are expressed in terms of the stress components. Some authors [77,78] propose instead to derive the resistance criteria by imposing an instability condition on the mathematical expression of the stress–strain relation. Although this approach is more correct from a theoretical point of view, it presents the drawback of being tied to the particular damage model used and therefore will not be considered here.

3.3.2. Modeling of the mechanical behavior of SiC//SiC composites

Regardless of the manufacturing process, the non-linear mechanical behavior of SiC//SiC composites depends on the interactions between the fibers and matrix during loading. These interactions lead at first to matrix microcracking, then to matrix/fiber decohesion, followed by the opening of microcracks, and finally to fibers (i.e. composite) failure. Specific behavioral models have been developed in the last years and one of them has recently been implemented in the finite element method code CASTEM2000 used for the thermo-mechanical analyses of the TAURO blanket [13]. Fig. 29 shows the results of tension test simulations at different loading directions, and compares them with those obtained assuming linear elastic behavior. Numerical values are referred to the 2D-CVI-Ceraset® composites. Two major stages characterize the mechanical behavior: in the first stage, cracks grow perpendicular to the loading direction and the in-plane mechanical behavior is isotropic; in the second stage, crack

![Fig. 29. Simulation of in- and off-axis incremental tension–compression test performed with the finite element code CASTEM2000 using a specific behavioral model for SiC//SiC composites.](image_url)
growth occurs perpendicular to the fibers and the mechanical behavior depends on the loading direction.

3.3.3. Resistance criteria

Numerous failure criteria have already been proposed for composite materials [79,80]. In their simplest form they consist of a number of expressions each of which is related by a separate stress/strain term. This allows for some physical insight into the mode of failure, but interactions between different terms are neglected. In order to take those into account the most commonly used are quadratic criteria, like the Tsai–Wu criterion [81], which have the form:

\[ F_i \sigma_i \sigma_j + F_j \sigma_j = 1 \quad (i, j \text{ summed}) \quad (1) \]

where \( F_i \) and \( F_j \) are strength parameters and \( \sigma_i \) and \( \sigma_j \) are the components of the stress tensor. However, combined stress tests are necessary to completely identify all the parameters, and these are not easy to perform. Moreover, quadratic criteria tend to be purely empirical, their purpose being to define a failure envelope by using a minimum number of test data.

The availability of the parameters needed to identify the criterion is an important issue when dealing with composites under constant development. The only two parameters \((\sigma_0^+ \text{ and } \sigma_0^-)\) are strength limits assumed for the TAURO blanket (based on Cerasep® N3-1) are reported in Section 2.2.2.

- The criterion used for plane stresses is a quadratic criterion derived from the Von Mises stress. The idea is to distinguish between tensile and compressive behavior on the basis of the spectral decomposition of the stress tensor. In terms of the invariants of this tensor the Von Mises criterion can be written as:

\[ \frac{1}{2} [3 \text{Tr}(\sigma_0) - \text{Tr}^2(\sigma_0)] \leq \sigma_0^2 \quad \text{or} \quad \frac{1}{2} [3 \text{Tr}(\sigma_0) - \text{Tr}^2(\sigma_0)] \leq 1 \quad (4) \]

where \( \sigma_0 \) is the maximum allowable stress (which is the same for compressive and tensile loading). We can then split the tensor \( \sigma \) into its positive and negative eigenvalues:

\[ \sigma^+ = \sigma^+ + \sigma^- \quad \sigma^+ \cdot \sigma^- = 0 \quad (5) \]

and modify inequality (4) as follows:

\[ \frac{1}{2} [3 \text{Tr}(\sigma_0^+) + 3 \text{Tr}(\sigma_0^-)] - \text{Tr}^2(\sigma_0^-) \leq 1 \quad (6) \]

or:

\[ \frac{(\sigma_0^+)^2}{(\sigma_0^-)^2} + \frac{(\sigma_0^-)^2}{(\sigma_0^+)^2} + \frac{(\sigma_0^+)^2}{(\sigma_0^-)^2} - \frac{\sigma_0^+ \sigma^-}{(\sigma_0^-)^2} \leq 1 \quad (7) \]

where \( \sigma_0^+ \) and \( \sigma_0^- \) are, respectively, the assumed limits under tensile and compressive loading. The only two parameters \((\sigma_0^+ \text{ and } \sigma_0^-)\) needed to identify this criterion can therefore be obtained by means of classic tension/compression tests. It must be stressed that the above criteria have been theoretically derived and their validation will require a large experimental campaign, which will have to be launched for the selected composites to be used in fusion blankets. Moreover, the effects of mechanical damage on the other composite physical properties (thermal and electrical conductivities, hermeticity, permeation rate, erosion rate, resistance to neutron irradiation etc.) are not yet known. In order to limit damage
effects within the composites the threshold of matrix microcracking saturation (beginning of fiber/matrix debonding) has been assumed as the maximum allowable tensile stress in the thermo-mechanical analyses of the TAURO blanket (about 145 MPa measured on 2D Cerasep\textsuperscript{®} composites). The actual failure limit (580 MPa) has been assumed instead as the maximum allowable compressive stress since no damage is observed under compression. The resulting failure envelope is shown in Fig. 30 and compared with those obtained with the Von Mises criterion assuming separately, $\sigma^+$ and $\sigma^-$ as limits.

- It is clear that the conventional primary and secondary stress limits do not apply for ceramics and that the above TAURO design criteria go a long way in providing more realistic stress criteria for SiC\textsubscript{f}/SiC design application. In the absence of tools enabling such sophisticated analysis, it is suggested that, as a very rough and conservative first order approximation, designers using conventional FEM analysis limit the total combined stress in plane to about 190 MPa. However, stresses through the thickness should be evaluated separately.

### 3.4. Lifetime

Lifetime is a very important parameter for fusion power plant studies as it has a direct effect on replacement costs and availability. However, it is very difficult to quantify and justify at this point in time as the database for modern SiC\textsubscript{f}/SiC composites is very limited. Many of the phenomena that could affect the lifetime of SiC\textsubscript{f}/SiC composites for a particular design have been identified in the previous sections and sub-sections. In some cases quantification of these phenomena has been limited to less-than-optimized SiC monoliths, fibers and composites. In other cases, these phenomena have been observed but not clearly understood from a mechanistic point of view.

Traditionally, the lifetime of structural components depends strongly on temperature, stress, chemical compatibility, thermal stability, radiation stability, transmutation products and dimensional stability. These parameters lead to time-dependent phenomena that determine failure boundaries, and hence, design lifetimes. Of all the phenomena identified in previous sections, two stand out as potential life-limiting mechanisms: compatibility with coolant (particularly Pb–17Li) and the effects of He transmutations. With regard to the compatibility issue, it is anticipated that pure SiC is thermodynamically stable in pure liquid Pb–17Li. However, oxide ceramics tend to be less stable in Pb–17Li. Thus, the compatibility of silicon-carbide matrix composites in Pb–17Li will be highly dependent on the oxygen level and the distribution within the material. The presence of silica on the surface of the composite, or even on the surface of the fibers, could have a profound effect on the compatibility. In general, compatibility is ensured by limiting the coolant/structure interface temperature and the exposure time, as well as the impurities. Compatibility data are essential to the establishment of both temperature limits and design lifetimes.
The helium transmutation rates are very high in silicon carbide composite blankets with a lead–lithium coolant/breeder. The transmuted He can have a significant effect on swelling (dimensional stability), hermeticity (coalescence of He bubbles on grain boundaries and interfaces), mechanical properties (weakening of joints and interfaces), and thermal conductivity (added resistance to heat flow). In addition to expanding the current programs studying the effects on implanted He, it is essential to conduct neutron-irradiation studies with concurrent He production in order to address potential life-limiting phenomena associated with He transmutation.

Other time-dependent phenomena that could affect lifetime are radiation-enhanced creep (dimensional stability), thermal creep (dimensional stability and failure) and thermal fatigue (failure). Although data are available for older fibers and composites, it is essential that these phenomena be quantified for the newer materials.

Given the state of the art of SiC/SiC composite development, it is premature to specify criteria for design lifetime at this point. The R&D effort should be developed in such a way that the beginning-of-life performance (based on unirradiated properties) satisfies design goals, that the early-in-life (one week to one month) performance satisfies design goals, and that the long term performance is adequate for attractive fusion power plant performance.

In the absence of firm guidelines, some studies have assumed certain limits. For example, the ARIES studies [7] have assumed a very rough estimate of 3% burnup (1.5 atom% He). Certainly this is an area where future R&D should be directed. No specific value is proposed for design studies at this time.

4. R&D plan

Although much has been accomplished through the international R&D effort over the last decade or so, a new impetus is required to push the R&D further in light of the results obtained so far and of the key issues identified through design studies for high performance SiC/SiC-based fusion power core.

Material R&D effort is continuing internationally and should be closely coordinated with the design effort. Current efforts are focused both on theoretical (modeling and analyses), experimental and manufacturing aspects in order to achieve the necessary SiC/SiC improvement and to establish its relevance for use as a structural material in fusion power core components.

In the EU, theoretical activities include the development of the TAURO Pb–17Li-based blanket conceptual design with the double objective of: (i) improving behavioral modeling and results interpretation; and (ii) supplying useful guidelines for the material development and characterization in order to relax some critical issues concerning the manufacturing of complex shapes. The development of a verification methodology for structural analysis is a mandatory action for any reactor design study involving ceramic matrix composites. Further activities will be carried out in order to identify a more refined model and a resistance criterion; in particular a new micromechanical damage model could be investigated by CEA in order to decrease the number of mechanical tests required for the identification of parameters describing the behavior of the composite.

The improvement of both the thermal conductivity and the stability of the thermo-mechanical properties during irradiation is a fundamental issue for SiC/SiC R&D: the continual progress in fiber quality, which led to the fabrication of almost stoichiometric fibers, indicates good progress towards reducing such a concern. However, a large effort has to be conducted for improving the matrix–fiber interface. Improvements of the SiC matrix processing to reduce the differences in performance with respect to the bulk CVD SiC seem more challenging as an R&D pursuit. New composites development is highlighted in the current R&D program. SNECMA has recently proposed a study for improvement of its composites based on: (i) the decreasing of porosity from <10% to <4% by means of a new matrix siliconized grade reaction giving SiC/SiC + Si composite; and (ii) the use of Hi-Nicalon S fibers with a new interphase + CVI matrix. New composites will be also manufactured by ENEA via polymer infiltration and pyrolysis by using advanced poly-
mer and stoichiometric fiber 3D preforms and performing the pyrolysis at higher temperatures in order to provide an almost crystalline β-SiC matrix. The capability of producing large-scale half-finished products and complex small mock-ups or large prototypes is well documented by SNECMA.

A number of irradiation tests are being planned internationally, as summarized in Table 9. For the irradiation testing scheduled at the High Flux Reactor at Petten, NL, the SiCf/SiC composites will be provided by the European Union, the United States and Japan in the framework of an IEA collaboration. The main properties to be investigated will be the bending strength and thermal diffusivity following irradiations at 900 and 1300 K temperature levels and up to a damage dose of 2.5 dpa. The possibility of some in-pile conductivity monitoring will be evaluated.

A study of the stability of SiCf/SiC composites in Pb–17Li is ongoing. The new exposure campaign will be carried out up to 6000 h in physico-chemical conditions approaching those representative of the TAUCRO blanket coolant (i.e. temperature of about 800 K, flowing LiPb with 0.5–1 m/s velocity). After exposure the samples will be characterized in order to investigate thermal stability and the degradation of thermo-mechanical properties.

In the US, the R&D plan has a primary goal of understanding the scientific basis for the behavior of SiCf/SiC composites and a secondary goal of applying this understanding to improve the performance of these materials. Examples of this approach are the following two tasks: (1) to develop a fundamental understanding of the radiation damage in monolithic SiC and SiC fibers with and without gaseous transmutations, i.e. with and without He; and (2) to improve the radiation stability in SiC fiber/interphase/matrix structures. The fundamental understanding will help identify the limits to these materials and to assist in the selection of optimized fiber/interphase/matrix combinations. The fundamental understanding will also underpin efforts to model the radiation response of SiCf/SiC composites. Fusion energy systems will require components larger than the current size of SiCf/SiC composites that can be fabricated. Therefore, the US program is also focused on development of low activation, radiation resistant joining methodology. The starting point for this effort is the joining methodology developed for non-nuclear programs with the US fusion program adapting these techniques to produce the desired properties. Thermal conductivity is a third topic that the US program is currently investigating. There is an effort to model radiation effects on the thermal conductivity of these materials and to develop engineering approaches to enhancing their thermal conductivity. Two smaller research efforts include: (1) demonstrating techniques for production of large components; and (2) development of a database on the high-temperature fracture strength, creep and thermal conductivity.

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Table 9

<table>
<thead>
<tr>
<th>SiC irradiation tests</th>
<th>Reactor</th>
<th>Temperature (°C)</th>
<th>Irradiation level (dpa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>12J and 14J (monolithic and composite: thermal conductivity, bend strength, dimensional change fiber strain and creep)</td>
<td>HFIR</td>
<td>500, 800</td>
<td>&lt;10</td>
</tr>
<tr>
<td>Mapping Elevated Temperature Swelling (METS) capsules (single crystals and Morton CVD)</td>
<td>HFIR</td>
<td>600–1500</td>
<td>1.5, 3.0</td>
</tr>
<tr>
<td>In planning stages. Will probably include composite thermal conductivity and fiber creep</td>
<td>HFR</td>
<td>600–1000</td>
<td>2.5</td>
</tr>
<tr>
<td>Composite thermal conductivity</td>
<td>ATR</td>
<td>300</td>
<td>2</td>
</tr>
<tr>
<td>Thermo-mechanical properties</td>
<td>JOYO</td>
<td>400–600</td>
<td>≤30</td>
</tr>
<tr>
<td>He effect through boron injection followed by irradiation</td>
<td>JMTR</td>
<td>500–1000</td>
<td>≈1</td>
</tr>
</tbody>
</table>

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The US R&D plan recognizes the importance of collaboration with the EU and Japanese fusion materials programs, of non-nuclear ceramic composite programs to provide basic material development and property input and of small business innovative research efforts. The goal of the US program is to address issues that are unique to fusion and rely on other sources of information for much of the fiber and composite development and database.

In Japan, the theoretical activities include the development of the DREAM and D-SSTR conceptual designs under JAERI and a new activity on He-cooled solid blanket design study as a part of the next Japan–US collaboration program from 2001 to 2007. The theory and modeling work is also recognized to be a very important part of the SiC/SiC R&D for fusion applications.

The development of structural analysis methodology and establishment of design codes with sufficient and appropriate safety margin are especially important for the utilization of SiC/SiC and ceramic matrix composites and would be carried out in collaboration by JAERI/NIFS and universities. The efforts to develop chemical and microstructural analysis methodologies and establishment of mechanical property evaluation methodologies with the emphasis on small specimen test technology (SSTT) for irradiated materials are important ongoing activities in universities and NRIM. The R&D efforts to develop simulation damage studies for 14-MeV neutrons utilizing electrons, ions and fission neutrons have been strengthened, together with the key technology development for IFMIF.

The improvement of both the thermal conductivity and the stability of the thermo-mechanical properties before and after irradiation has been an important issue in the Japanese program, where significant fiber property improvements have been accomplished and are still ongoing. These improvements include achieving near-stoichiometry composition with higher crystallinity and exploring effects of metallic element additions, in fibers such as Hi-Nicalon type-S, Tyranno-SA and others. Efforts have also been underway for improving matrix–fiber interface and for improving the SiC matrix in order to improve total performance of SiC/SiC composites. These efforts for fusion application are based on non-fusion R&D activities in Japan, such as the AMG program, the Sunshine project and others. CREST-ACE program and the current Japan–USA collaborative program, JUPITER, are the most active efforts in trying to develop SiC/SiC for fusion application. Further development of composite fabrication process is in progress including: (1) multiple-interphase + CVI matrix, (2) PIP (polymer infiltration and pyrolysis) using new polymers and their blends, (3) RS (reaction sintering, or melt infiltration) using filler powders, (4) HP (hot pressing), and (5) a combined process of (1)–(4). These process developments are in progress at Kyoto University, Osaka Prefecture University, NRIM, JAERI, Tokyo Institute of Technology and others.

Kyoto University/Ube has produced tubes and cone shaped-components for hermeticity/vacuum properties measurement (Hokkaido/Kyoto), tritium inventory test (Kyushu/Kyoto), UNICEX (unit cell experiment to be done by Kyoto/UCLA as a part of the next Japan–USA collaboration program, JUPITER-II) and others. This effort has just started and will be continued extensively for three to four years.

As an ISTC program, a very preliminary study on in-reactor creep of SiC/SiC is ongoing between Japan and the Russian Federation. The new Japan–US collaborative program, JUPITER-II, will be initiated on April 1, 2001, under which the following tasks will be investigated:

**Task 2: High-temperature gas-cooled blanket**

- Sub-task 2-1: Pre-irradiation studies of key materials system issues
  - SiC hermetic joining, coatings, thermal conductivity improvement, high temperature corrosion, and compatibility with Be breeders
  - SiC fabrication R&D and materials supply
  - SiC materials system thermal–mechanical interactions
- Sub-task 2-2: Irradiation testing of materials systems
– SiC capsule irradiations

As part of the domestic programs, fast neutron irradiation in JOYO and mixed-spectrum neutron irradiation in JMTR are in progress. The main properties to be investigated will be the bending strength, thermal diffusivity, tensile strength, fatigue properties, and creep properties following an irradiation at a temperature of up to 1300 K and a damage dose of up to 5 dpa. The possibility of some in-pile conductivity monitoring will be evaluated. A new high resolution HVEM + ion accelerator at CARET, Hokkaido University, will provide in-situ observation results of radiation damage in SiCf/SiC composites. A new multiple beam irradiation test facility at Kyoto University (DuET Facility) will provide data from 4 to 1600 K (in vacuum and in gaseous environments) under precisely controlled conditions for systematic and accurate experiments for establishing radiation damage process integration.

Studies of the stability of SiCf/SiC composites in Pb–17Li, Li and Flibe are ongoing. This activity will be expanded as a part of the upcoming JUPITER-II program under the collaboration with INEEL.

5. Conclusions

The use of SiCf/SiC composite as a structural material in fusion reactors is attractive based on its low induced radioactivity and afterheat. It has been considered in several power plant studies, most recently in the TAURO, DREAM, and ARIES-AT design studies. From these studies, several issues have been identified for the SiCf/SiC material. These include the cost of fabrication, the joining methods, and the factors limiting the range of operation and performance, in particular the rather low thermal conductivity at high temperature and under irradiation, and the maximum allowable operating temperature (based on strength degradation and/or compatibility). These have identified the progress required in developing the new generation SiCf/SiC in order to make it an attractive high performance fusion power core material with safety advantages.

Based on the available results and a reasonable extrapolation in improving SiCf/SiC, in particular the goal of achieving stoichiometry (minimizing impurities), the SiCf/SiC parameters and properties shown in Table 5 are recommended for present design analysis of SiC/SiC-based power plant for the long term (20–30 years in the future, or more). This is in agreement with the suggestion emerging from the last International Town Meeting on SiC/SiC Design and Material Issues for Fusion Systems [26].

Acknowledgements

This paper resulted in great part from very informative discussions at the ‘International Town Meeting on SiC/SiC Design and Material Issues for Fusion Systems’, which was held at Oak Ridge National Laboratory on January 18–19, 2000 [26]. The meeting was very successful in achieving its main objective of bringing together the SiC/SiC design and material communities from Japan, the EU and the US to exchange information, identify the design-related critical issues, discuss them in the light of the latest material R&D results, and provide guidelines to help focus future effort. The authors gladly acknowledge the contributions of all the meeting participants.

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