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PROPERTIES OF SILICON CARBIDE
RELEVANT TO ITS PERFORMANCE
AS A COATING FOR HIGH TEMPERATURE
NUCLEAR FUEL PARTICLES

by

R. J. Price, T. D. Gulden, and J. L. Kaae

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PROPERTIES OF SILICON CARBIDE RELEVANT TO ITS PERFORMANCE AS A COATING FOR HIGH TEMPERATURE NUCLEAR FUEL PARTICLES

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INTRODUCTION

Pyrolytic silicon carbide has certain advantages as a coating material for high-temperature nuclear fuel particles. This summary outlines the present state of knowledge of the physical properties of silicon carbide which are important in determining the service behavior of coated nuclear fuel particles, and also presents some results obtained from a mathematical model which describes the stress state in a composite coating of pyrolytic carbon and silicon carbide on a fuel particle as a function of burnup and neutron irradiation.

DIMENSIONAL STABILITY UNDER IRRADIATION

Both α and β silicon carbide expand when irradiated with fast neutrons. The expansion is virtually isotropic, even for crystals of α (hexagonal) silicon carbide. The expansion saturates at low doses ($\sim 3 \times 10^{20}$ nvt, fast).⁽⁷⁾ The linear expansion at saturation decreases with increasing irradiation temperature, falling from $\sim 1\%$ at room temperature to less than 0.1% at 1000°C . The saturation expansion is shown as a function of irradiation temperature in Figure 1, which incorporates data from several published sources in addition to results obtained at GGA. The graph is restricted to neutron exposures high enough for saturation to have occurred. The highest exposure points are those for pyrolytic β -silicon carbide irradiated in P-13-H at $\sim 900^\circ\text{C}$ to 4.2×10^{21} nvt ($E > 0.18$ Mev).⁽²⁾ The expansion decreases steadily with increasing irradiation temperature. There does not seem to be any systematic dependence on the structure or density of the material.

The lattice parameter of irradiated silicon carbide increases by the same fraction as the macroscopic linear expansion, even at the highest irradiation temperatures studied. The x-ray line breadths show a slight increase which corresponds to a small increase in internal elastic strain ($\sim 3 \times 10^{-4}$ after 3.8×10^{21} nvt (fast) at 650°C and $\sim 8 \times 10^{-4}$ after 4.2×10^{21} nvt (fast) at 900°C , compared with an as-deposited strain of $\sim 10^{-3}$).

The rapid saturation of expansion with increasing dose shows that equilibrium is soon established between the rate of displacement of atoms and the rate of recombination of interstitials and vacancies. The temperature dependence of the saturation expansion shows the recombination to be thermally activated. A model for irradiation-induced defect production and annihilation proposed by Hickman and Pryor⁽⁸⁾ predicts that the saturation expansion should be higher for faster dose rates, and that saturation should be achieved more rapidly at higher irradiation temperatures. However, the data in Figure 1 were obtained from a variety of irradiation facilities with dose rates ranging from $\sim 2 \times 10^{15}$ n/cm²-sec, in the Dounreay Materials Testing Reactor (data from Thorn et al) to $\sim 10^{14}$ n/cm²-sec, and the results do not show any apparent dependence on dose rate within the scatter of the data. These dose rates are considerably higher than those anticipated in an HTGR ($\sim 5 \times 10^{13}$ n/cm²-sec), and therefore, if there were a dose-rate dependence, the effect would be to reduce the expansions occurring in service to below the values shown in Figure 1.

A comparison of the expansion occurring in silicon carbide with that occurring in several other ceramic materials is given in Figure 2 for an irradiation temperature of $\sim 150^\circ\text{C}$. Silicon carbide resembles aluminum oxide and magnesium oxide in that the expansion shows saturation with increasing exposure. The silicon carbide expands faster and saturates at a higher level than the alumina and magnesia. In contrast, the beryllia and graphite continue to expand and show no sign of saturation.

THERMAL CONDUCTIVITY

UNIRRADIATED

The thermal conductivity of silicon carbide is sensitive to the purity and porosity of the material. At room temperature the conductivity lies in the range of 0.1 to 0.5 cal/cm-sec- $^\circ\text{C}$.⁽¹¹⁾ The conductivity decreases with increasing measurement temperature. At 1000°C the conductivity is typically one fifth of its room temperature value.^(11, 12) The room temperature conductivity of pyrolytic silicon carbide prepared at 1400°C at GGA is about 0.1 cal/cm-sec- $^\circ\text{C}$. The present equipment does not permit measurements at high temperatures.

IRRADIATED

Measurements on low-density commercial silicon carbide⁽⁷⁾ show that the thermal conductivity falls on irradiation to a level independent of exposure (beyond 2×10^{20} nvt)

but dependent on irradiation temperature. The room temperature conductivity decreases by a factor of 16 after irradiation at 250°C, 11.5 at 475°C, and 8.8 at 700°C. Extrapolation suggests a decrease by a factor of about 6 at 1000°C.

There is no information on the effect of measurement temperature on the conductivity of irradiated silicon carbide. Thermal diffusivity samples of GGA pyrolytic silicon carbide are currently under irradiation at ~1000°C to ~4 x 10²¹ nvt (E > 0.18 Mev) and at ~500°C to ~3 x 10²¹ nvt (E > 0.18 Mev). Post-irradiation measurements will be made at room temperature. Higher temperature measurements will be made if suitably modified equipment can be made available.

MECHANICAL PROPERTIES OF SILICON CARBIDE

The mechanical properties of materials in general and of ceramic materials in particular are often strongly dependent on grain size. The grain size of vapor deposited SiC may be varied between rather wide limits by varying the deposition temperature and the partial pressure of methyltrichlorosilane. Thus, it is of interest to determine the variation of the mechanical properties of SiC with grain size in order to optimize the deposition conditions with respect to coating strength. It is also important to obtain data at the temperatures of interest since grain boundary shear and creep may make significant contributions to the elevated temperature properties.

ROOM TEMPERATURE MECHANICAL PROPERTIES OF SiC

The fracture stresses of vapor deposited SiC with mean grain diameters in the range of less than 1 micron to 20 microns have been measured in four point bending. Preliminary results suggest that the dependence of fracture stress on the grain size follows a Petch relation

$$\sigma_f = \sigma_y + k_y D^{-1/2}$$

Where σ_y and k_y are constants and D is the average grain diameter. The data are shown in Figure 3. The average fracture stress of SiC with a one micron grain size is about 125,000 psi and that of the 20 micron grain size SiC is about 107,000 psi.

This dependence of fracture stress on grain size is consistent with published results on the fracture of polycrystalline oxide ceramics in this range of grain sizes.⁽¹³⁾ The dependence of fracture stress on the inverse square root of the grain size is generally explained in terms of the presence of built-in cracks

which extend completely across individual grains, but which do not penetrate the grain boundaries. The presence of a grain size independent term in the equation describing the fracture of SiC implies a fracture mechanism which involves yielding as opposed to the classically brittle case described by the Griffith-Orowan equation. It is not clear whether or not the yielding requires dislocation motion, which must be very difficult in the case of SiC at room temperature.

ELEVATED TEMPERATURE MECHANICAL PROPERTIES OF SiC

Equipment for measuring the mechanical properties of SiC in an inert atmosphere at temperatures up to about 1500°C has been designed and is presently being constructed. No results have yet been obtained on this phase of the program. However, some inferences on the high temperature mechanical properties of SiC may be made from published results on similar materials.

The mechanical properties of vapor deposited SiC might be expected to be somewhat analogous to those of polycrystalline oxide ceramics, since in both cases the amount of dislocation motion that can occur will be very limited except at extremely high temperatures. In the case of oxides the trend of increasing strength with decreasing grain size is often reversed at high temperatures. Specimens with intermediate grain sizes (about 20 microns) are stronger than those with a very fine grain size. This reversal can generally be correlated with purity, occurring at lower temperatures in less pure materials, and has been attributed to the onset of grain boundary shearing.⁽¹³⁾ The purity of vapor deposited SiC appears to be quite high and this effect may not be important at the temperatures of interest. It is thought that fine grained material will probably continue to be the strongest at temperatures up to about 1200°C.

Some information is available which indicates that the bend strength of relatively impure hot pressed SiC, of undefined grain size, changes very little between room temperature and 1200°C.⁽¹⁴⁾ Most of the published data on mechanical properties of SiC were obtained on poorly characterized material and are of little or no use in predicting the behavior of vapor deposited SiC.

An additional factor which may be important to the high temperature mechanical properties of SiC is creep. A systematic investigation of the high temperature (1900°C to 2200°C) creep of hot pressed hexagonal SiC has been reported by Farnsworth and Coble.⁽¹⁵⁾ The behavior was found to be phenomenologically

consistent with a mechanism of creep by stress-induced flow of vacancies (Nabarro-Herring mechanism). The temperature dependence of the process was thought to be that for grain boundary diffusion. This type of creep behavior is generally observed in ceramic materials which do not have enough slip systems to deform by dislocation motion. It is likely that grain boundary shear also makes a significant contribution to the creep of these materials.⁽¹⁶⁾ Extrapolation of the Farnsworth and Coble data to 1000°C and to the grain size typical of vapor deposited SiC (1 micron) leads to the prediction of a significant creep rate for conditions that will be encountered in practice by SiC coatings. The prediction of high creep rates does not seem to be supported by the experiments on irradiation creep reported in a later section⁽²⁾. The experimental techniques employed by Farnsworth and Coble are open to some question and the extrapolation is over too wide a range to be trusted with any degree of confidence. Considerable additional work is needed in this area before the contribution of creep to the high temperature mechanical properties is completely understood.

The ductility of most ceramic materials under creep conditions is severely limited, with failure occurring by grain boundary separation. The creep strain to failure may well set a very low limit on the creep rate that can be tolerated in practice. The creep rate may readily be controlled by varying the grain size. A dependence of creep rate on the inverse cube or square of the grain size is generally observed in ceramic materials. Therefore, an increase in grain size from 1 micron to 10 microns would be expected to reduce the creep rate by a factor of 100 or 1000.

EFFECT OF IRRADIATION ON MECHANICAL PROPERTIES

Both the modulus of rupture and Young's modulus of self-bonded low-density β -silicon carbide have been reported to decrease by 5 - 10% after irradiation to 1.8×10^{21} nvt (fast) at temperatures between 250°C and 700°C.⁽⁷⁾ Room-temperature tests on pyrolytic silicon carbide deposited at 1400°C showed that there was no change in either elastic modulus or modulus of rupture after irradiation at 650°C and 1030°C to $2.7 - 2.8 \times 10^{21}$ nvt ($E > 0.18$ Mev).⁽¹⁾ Material annealed for 20 minutes at 2000°C after deposition (corresponding to the deposition of an overcoat of high-temperature pyrocarbon) behaved similarly. The data are shown in Table 1. The samples were not lapped before testing in either the irradiated

TABLE I
CHANGE IN MECHANICAL PROPERTIES OF SILICON CARBIDE
STRIPS IRRADIATED IN P-13-F CAPSULE

Sample Number	Condition ^(a)	Cell Number	Temp (°C)	Exposure (10 ²¹ nvt, E > 0.18 MeV)	Modulus of Rupture (10 ³ psi)		Young's Modulus (10 ⁶ psi)	
					Individual	Mean ^(b)	Individual	Mean ^(b)
---	D	Unirradiated	---	0	---	92±20	--	50±2
---	A				---	116±17	--	43±4
D-1000-1	D	1	650	2.8	110	117±15	49	49±3
D-1000-2	D				134		52	
D-1000-3	D				108		46	
A-1000-1	A				118	120±3	43	42±1
A-1000-3	A				122		41	
D-1400-1	D	2	1030	2.8	106	89±25	52	52±1
D-1400-2	D				71		51	
A-1400-1	A				114	121±10	45	44±1
A-1400-3	A	3	1025	2.7	128		43	

^(a)D = as deposited at 1400°C; A = annealed at 2000°C for 20 min

^(b)With estimated standard deviation

or unirradiated condition, which accounts for the low apparent values of elastic modulus and strength, and also for the slight difference in properties between the as-deposited samples and the annealed samples.

The x-ray line-broadening measurements described earlier indicate that the internal stresses generated on irradiation at 650°C and 900°C are very small. In general, such stresses are lower than the as-deposited stress.

IRRADIATION CREEP

"Creep" samples of pyrolytic silicon carbide were irradiated in capsule P-13-H at 650°C to 3.8×10^{21} nvt ($E > 0.18$ Mev) in cell 1, and at 900°C to 4.2×10^{21} nvt ($E > 0.18$ Mev) in cell 3. The samples consisted of graphite substrate slabs over-coated with silicon carbide. The substrate slabs, which measured about 6mm x 1mm x 1mm, were cut with their long axis at 40°, 30° or 0° to the hot-working axis of a block of hot-worked graphite in order to apply one of three different strain rates to the silicon carbide coating. The radiation-induced expansion of the graphite substrate resulted in the silicon carbide being stretched in uniaxial tension at an approximately constant strain rate.

The strains in the creep samples were measured from the change in spacing of pairs of fiducial marks. The results are shown in Table II, together with the values for the unrestrained controls. With the exception of two low strain-rate samples in the low-temperature cell, the expansion of the substrate graphites caused cracking in the strained silicon carbide coatings. One of the two unbroken samples received no significant creep strain, but the other apparently survived an applied creep strain of 0.3%.

The creep rate of many materials under neutron irradiation is a linear function of stress:

$$\frac{\xi}{\gamma} = K\sigma$$

Where ξ is the creep strain, γ is the neutron exposure, K is the creep constant and σ is the stress. Assuming the same law applies to silicon carbide and neglecting transient creep, the present results may be used to place rough limits on the value for K . This approach assumes that the samples failed on reaching their fracture stress, rather than on reaching a limiting creep strain. The fracture stress of the silicon carbide, σ_f , may be taken as one half the modulus of rupture

TABLE II

STRAINS IN UNRESTRAINED AND CREEP SAMPLES OF SILICON
CARBIDE IRRADIATED IN P-13-H CAPSULE

Cell	Temp (°C)	Exposure (nvt, $E > 0.18$ MeV)	Sample Number	Strain Rate (%/10 ²¹ nvt)	Change in Length (%)	Survived or Failed	Creep Strain (%)
1	650	3.8×10^{21}	SU-1	0	+0.25	---	
1			SU-5	0	+0.33	---	
1			SR-1-1	0	+0.28	Survived	0.00
1			SR-2-1	0.12	+0.67	Hairline cracks	<0.37
1			SR-3-1	0.5	+1.9	Cracked	<1.6
1			SU-2	0	+0.30	---	
1			SU-6	0	+0.33	---	
1			SR-2-2	0.09	+0.60	Survived	0.30
1			SR-3-2	0.3	+1.2	Cracked	<0.9
3			SU-3	0	0.00	---	
3	900	4.2×10^{21}	SU-7	0	+0.07	---	
3			SR-1-2	0.19	+0.81	Cracked	<0.74
3			SR-2-3	0.5	+2.0	Cracked	<2.0
3			SR-3-3	1.2	+4.6	Cracked	<4.5
3			SU-8	0	+0.15	---	
3			SR-1-3	0.24	+1.0	Cracked	<0.92
3			SR-2-4	1.2	+4.5	Cracked	<4.4
3			SR-3-4	2.1	+8.1	Cracked	<8.1

or $\sim 60,000$ psi. Thus for the surviving sample

$$K > \frac{\dot{\epsilon}}{\gamma \sigma_f}$$

and for the failed samples

$$K < \frac{\dot{\epsilon}}{\gamma \sigma_f}$$

Using the data in Table II, at 650°C K is estimated to be between 1.5 and $2 \times 10^{-9}(\text{psi})^{-1} (10^{20} \text{ nvt})^{-1}$, and to be less than $3 \times 10^{-9}(\text{psi})^{-1} (10^{20} \text{ nvt})^{-1}$ at 900°C . These values are two orders of magnitude lower than those found in pyrolytic carbons.

Further irradiation creep experiments are planned for a forthcoming capsule. Different graphite substrates will be used in an attempt to impose lower strain rates than those occurring in the experiments described above. It is hoped that these experiments will enable closer limits to be placed on the estimates of the creep constant.

STRESS STATE IN PyC-SiC COATINGS

The stress state developed in a composite pyrocarbon and silicon carbide coating on a fuel particle with increasing burnup depends on the irradiation stability and on the mechanical and physical properties of the silicon carbide and pyrocarbon as well as on the fission-gas pressure in the interior of the particle. In order to understand the interaction of these factors, the stress state in a composite coating was determined as a function of the neutron dose by means of a mathematical model which attempted to take into account all of the major elements which contributed to the stresses. The model considered:

1. A spherical particle with the silicon carbide layer inside of the pyrocarbon layer.
2. A continual pressure increase inside of the composite coating.
3. Stresses induced by the difference between the thermal expansion of silicon carbide and that of pyrocarbon.
4. A constant temperature throughout the buffer, the silicon carbide and the pyrocarbon layers.

5. Radiation-induced dimensional changes in the pyrocarbon.
6. Radiation-induced creep in the pyrocarbon.
7. A complete mechanical stability of silicon carbide, i.e., no radiation-induced dimensional changes and no radiation-induced or thermal creep.

A particle design with the silicon-carbide layer inside of the pyrocarbon layer seemed to be superior to a design with interior pyrocarbon and exterior silicon carbide since in the latter case a tensile stress would be imposed on the interface between the pyrocarbon and the silicon carbide. Such a tensile stress might cause the pyrocarbon to break away from the silicon carbide and this could be detrimental in terms of the level of internal pressure which would cause fracture of the coating. In addition, since it is desired to limit the amount of silicon carbide in the particle, with the same mass of material it would be possible to use a thicker layer of silicon carbide if the silicon carbide was inside rather than outside of the pyrocarbon.

The pressure increase inside of the composite shell was calculated in a fashion similar to that used by Schwartz.⁽¹⁷⁾ There is some uncertainty in this calculation, but the answer should yield conservative values for tensile fracture stresses in the particle. Stresses caused by the anisotropic dimensional changes in the pyrocarbon were calculated from the equations developed by Prados and Scott,⁽¹⁸⁾ and creep in the pyrocarbon was considered in the fashion described by Mendelson et al.⁽¹⁹⁾ A short description of the method of considering the interaction of the concentric hollow spheres has been described previously.⁽²⁰⁾

As is evident from Table IV, silicon carbide appears to be much more stable under irradiation than pyrocarbon, the radiation-induced dimensional changes seeming to be at least an order of magnitude less than those of pyrocarbon and the creep rate during irradiation at 900°C seeming to be about two orders of magnitude less than that of pyrocarbon. Since in the model the radiation and creep-induced dimensional changes of the silicon carbide must be superimposed on those of the pyrocarbon and since they are much smaller than those in the pyrocarbon, the assumption of stability of the silicon carbide in the model appears to be justified.

Although the investigation of the influence of changing the particle dimensions and of variation of the mechanical properties of the coatings on the results obtained from the model is not complete, it appears that in every case the

silicon carbide is at least initially subject to tangential compressive stresses. These compressive stresses in the silicon carbide vary widely with particle dimensions and properties; however, most important, it appears that with reasonable dimensions the silicon carbide layer can remain in compression at the end of the life of the particle, thus minimizing the chances of failure in the relatively brittle SiC layer. The tangential stresses in the pyrocarbon are always tensile. Depending on the creep constant and on the anisotropic dimensional changes assumed for the pyrocarbon, the stresses in the pyrocarbon either go through a maximum or reach approximate steady state values with increasing burnup or neutron dose. For example, for a composite particle with the dimensions and mechanical properties shown in Table III and with the anisotropic dimensional changes for pyrocarbon shown in Figure 4, the model predicts that the stresses will vary with neutron dose in the fashion shown in Figure 5.

The foregoing analysis is an initial approach which is believed to contain all of the most important elements of the problem. Additional work is planned to investigate the effects of variations in particle dimensions and variable coating properties. Another variable of unknown significance is that of deviation from spherical shape, which is being considered for later analysis.

TABLE III

Dimensions of the Assumed Fuel Particle

Core Diameter = 250 microns

Buffer Thickness = 35 microns - Buffer Density = 1.2 gm/cc

SiC Thickness = 15 microns

Pyrocarbon Thickness = 60 microns

The core was assumed to consist of 75% thorium carbide and 25% uranium carbide.

Burnup was assumed to occur linearly with neutron flux and was assumed to be 20% FIMA at 7.10^{21} nvt.

Physical and Mechanical Properties of the Assumed Particle

Pyrocarbon

$E = 4 \times 10^6$ psi $\mu = .33$ $\alpha = 5.0 \times 10^{-6} (^{\circ}\text{C})^{-1}$

Transient radiation-induced creep constant = $2.5 \times 10^{-27} (\text{psi} \times \text{nvt})^{-1}$

Steady state radiation-induced creep constant = $1.0 \times 10^{-27} (\text{psi} \times \text{nvt})^{-1}$

Transition between transient and steady state at 1.10^{20} nvt.

Silicon Carbide

$E = 60 \times 10^6$ psi $\mu = .33$ $\alpha = 4.5 \times 10^{-6} (^{\circ}\text{C})^{-1}$

Temperature = 900°C

CONCLUSIONS

A summary of some of the more important properties of silicon carbide is shown in Table IV, together with the corresponding properties of pyrolytic carbon for comparison.

The radiation-induced expansion of silicon carbide is isotropic and rapidly saturates at a low level ($\sim 0.1\%$ at 1000°C). By contrast, pyrolytic carbon undergoes large anisotropic dimensional changes which increase with increasing exposure (see Figures 2 and 4). The thermal conductivity of both materials is adequate for a coating on fuel particles operating at normal burnup rates. The thermal expansion coefficient of SiC is compatible with that of isotropic pyrolytic carbon. The mechanical strength of silicon carbide is over twice that of pyrolytic carbon and present results do not indicate any significant reduction in strength on irradiation.

Silicon carbide has a very high elastic modulus (60×10^6 psi) as compared to $\sim 4 \times 10^6$ psi for irradiated pyrolytic carbon, and is extremely brittle. Present results suggest that silicon carbide will remain much more brittle than pyrolytic carbon under irradiation at $\sim 1000^{\circ}\text{C}$. Carbon undergoes sufficient irradiation creep to provide a useful reduction in stresses, whereas current indications are that silicon carbide does not. It is possible that future work will clarify the creep behavior of silicon carbide and indicate how the microstructure may be chosen to optimize the creep properties.

Two approaches to circumvent the problem of brittleness of the silicon carbide coating materials are possible. The first would be to use a very large safety factor in designing the particle coating and to always make the coatings thick enough for the stresses to be limited to a small fraction of the fracture stress. The choice of a safety factor would depend on the statistics of the relevant parameters for large batches of particles, and the fraction of failed coatings that could be tolerated. The second approach is outlined in the last section of this report and involves a particle where a diffusion barrier-layer of silicon carbide is surrounded by a thicker outer layer of pyrolytic carbon. The shrinkage of the carbon would be relied on to keep the silicon carbide in compression for most of the life of the fuel. Such an approach would require good knowledge and tight control of the properties of the pyrocarbon coating.

TABLE IV

COMPARISON OF THE PROPERTIES OF SILICON CARBIDE AND PYROLYTIC CARBON

Property	β -Silicon Carbide	Pyrolytic Carbon	
		High-temperature isotropic	Low-temperature isotropic
Crystal Structure	Cubic	Turbostatic (faulted hexagonal, $c/a = 2.8$)	
Radiation-induced dimensional change rates at $\sim 1000^\circ\text{C}$	+ 0.1% (isotropic) Saturates	Perpendicular: + 2.5%/10 ²¹ nvt Parallel: - 2.0%/10 ²¹ nvt No saturation	Perpendicular: + 3%/10 ²¹ nvt (a) Parallel: - 2%/10 ²¹ nvt (a) No saturation
Room temperature thermal conductivity:- Unirradiated	0.1 cal/cm-sec- $^\circ\text{C}$	3×10^{-2} cal/cm-sec- $^\circ\text{C}$	10^{-2} cal/cm-sec- $^\circ\text{C}$
- Irradiated (to $\sim 2.5 \times 10^{21}$ nvt at $\sim 1000^\circ\text{C}$)	(b)	2×10^{-2} cal/cm-sec- $^\circ\text{C}$	(b)
Thermal expansion coefficient (at 400°C)	$4.5 \times 10^{-6} \text{ }^\circ\text{C}^{-1}$	$5 \times 10^{-6} \text{ }^\circ\text{C}^{-1}$	$6 \times 10^{-6} \text{ }^\circ\text{C}^{-1}$
Young's modulus - Unirradiated	60×10^6 psi	2×10^6 psi	5×10^6 psi
Irradiated (to $\sim 2.5 \times 10^{21}$ nvt at $\sim 1000^\circ\text{C}$)	60×10^6 psi	4×10^6 psi	(b)
Modulus of rupture - Unirradiated	120×10^3 psi	35×10^3 psi	50×10^3 psi
Irradiated (to $\sim 2.5 \times 10^{21}$ nvt at $\sim 1000^\circ\text{C}$)	120×10^3 psi	50×10^3 psi	(b)
Irradiation creep constant at $\sim 1000^\circ\text{C}$	$< 3 \times 10^{-29} (\text{psi})^{-1} (\text{nvt})^{-1} \text{ (a)}$	$1-2 \times 10^{-27} (\text{psi})^{-1} (\text{nvt})^{-1}$	$< 10^{-27} (\text{psi})^{-1} (\text{nvt})^{-1} \text{ (a)}$

(a.) Preliminary values

(b) No data available

Fig. 1. Irradiation-induced expansion in silicon carbide

Fig. 2. Radiation-induced expansion of various ceramic materials at 150°C

Fig. 3. Petch plot of preliminary silicon carbide fracture data at room temperature

Fig. 4. Radiation-induced dimensional changes in pyrolytic carbon. $\rho_0 = 1.95$ gm/cc; initial BAF ≤ 1.2 (Courtesy of J. Bokros, A. Schwartz, and R. Dunlap)

Fig. 5. Stresses in a Pyc. -SiC coating assuming dimensions and properties shown in Table 3

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