Fracture Strength of Silicon Carbide Layers of TRISO Particles Coated at Various Temperatures

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1. Introduction

TRISO-coated fuel particles for high-temperature gas-cooled reactors consist of UO_2 microspheres coated with layers of porous pyrolytic carbon (porous PyC), inner dense PyC (IPyC), silicon carbide (SiC), and outer dense PyC (OPyC). Among the TRISO coating layers, the SiC layer is particularly important because it acts as a diffusion barrier to gaseous/metallic fission products and as a miniaturized pressure vessel of the particle. In order to insure the integrity of the SiC layer after a fabrication and in use, the microstructure, mechanical properties, and chemical composition of the SiC layer should be properly controlled [1].

For a uniform coating of the microspherical particles, the TRISO coating is performed using a fluidized-bed chemical vapor deposition (FBCVD) method. Characteristics of the coating layer depend largely on the FBCVD conditions such as the gas flow rate, concentration of the coating gas, coating temperature, etc [2,3]. Among the deposition parameters the coating temperature is particularly important for determining the properties of the SiC layer. In this study, we investigated the effect of the coating temperature on the fracture strength of the SiC layer while fixing the other deposition parameters.

2. Experimental Procedure

Coatings of the TRISO particles were conducted using ZrO_2 kernels in a FBCVD reactor. In this work, a graphite tube of 25 mm inner diameter with an inlet nozzle of 3 mm at the base of a 60° cone was used as a coating bed. Input gases for the depositions of the buffer, IPyC (OPyC) and SiC were C_2H_2/Ar, C_2H_2/C_3H_6/Ar and CH_3SiCl_3 (MTS)/H_2/Ar, respectively. For the deposition of the SiC layer, the coating temperature was varied between 1400° and 1600°C at a constant gas flow rate and an input gas ratio ((Ar+H_2)/MTS) of 100. Coating thicknesses of the SiC layers were controlled in the range of 34–40 μm by varying the coating times. All the TRISO layers were continuously coated without unloading the particles after each coating step.

The fracture strength of the SiC layer was measured through a crush test of a hemispherical shell specimen. The TRISO-coated particles mounted in an epoxy resin were ground close to the equatorial plane of the particles and the SiC hemispherical shell specimens were collected after burning off the IPyC and OPyC layers. The hemispherical shell specimen, laid on an alumina plate, was loaded by an upper alumina rod. The crosshead speed used was 0.1 mm/min. The fracture strength of the SiC hemispherical shell specimen was correlated with the critical load for the cracking of the shell. A finite element analysis was used to derive a semi-empirical equation for the strength calculation [4]. A statistical analysis of the fracture strength was also performed by the two-parameter Weibull analysis. The number of samples used for the Weibull analysis was more than 80.

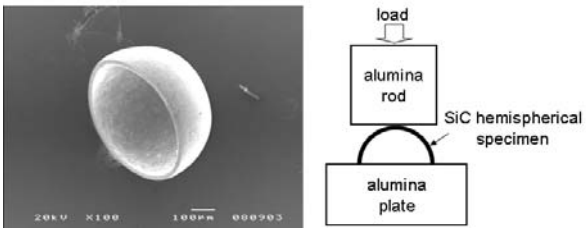

Fig. 1. SEM image of the SiC hemispherical shell specimen and the configuration of the crush test.

3. Results and Discussion

Fig. 1 shows a typical micrograph of the SiC hemispherical shell specimen after burning off the PyC layers and the configuration of the fracture strength test. The fracture was initiated at the inner surface of the hemispherical shell beneath the loading point. Weibull plots and a summary of the statistical parameters for the

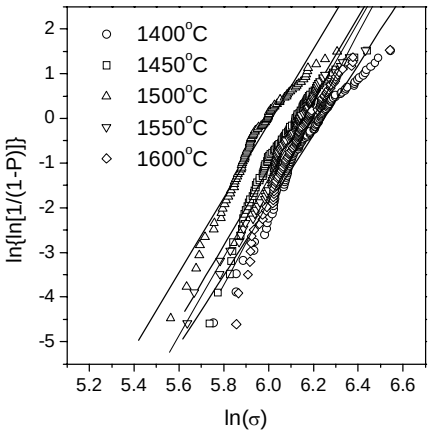

Fig. 2. Weibull plots for the fracture strengths of the SiC layers coated at various temperatures.

Table 1 Summary of the fracture strength tests of the SiC hemispherical shell specimens

Coating temperature (°C)	Sample size	Mean strength σ_m (MPa)	Coefficient of variation (%)	Weibull modulus m
1400	97	517	14.2	7.9
1450	98	464	13.1	8.3
1500	87	409	13.4	8.3
1550	98	471	12.5	8.8
1600	100	491	12.0	9.4

fracture strengths measured by the crush test of the hemispherical shell specimens are presented in Fig. 2 and Table 1, respectively. The mean strengths do not show a wide range of variation with varying coating temperatures although the 1500°C specimen has a slightly lower strength. Weibull moduli also reveal similar values between 7.9 and 9.4.

In our previous studies [3,5], the pore size and porosity of the SiC layer increased with increasing coating temperature, especially at 1550° and 1600°C. An appreciable degradation of the nano-indentation hardness and elastic modulus appeared for the SiC layers coated at 1550° and 1600°C, which could be attributed to the increased porosity of the specimens and partly to the existence of free carbon for the 1600°C specimen. Therefore, the results of the fracture strength tests are not correlated with the nano-indentation results in our previous studies and there is no clear dependency of the fracture strength on the coating temperature. These results can be interpreted as follows. The fracture of the SiC hemispherical shell specimen is initiated at the inner surface with a high roughness of several micrometers in size, which originates from the uneven outer surface of the IPyC layer, and the roughness is on a larger scale than the pore size in the SiC layer as shown in Fig. 3. Therefore, it is likely that the fracture strength is dominated by the roughness of the SiC inner surface, thus overshadowing the effects of the internal porosity and microstructural differences on the fracture strength.

Another aspect that one has to keep in mind, however, is that the roughness of the IPyC/SiC interface can be beneficial for increasing the debonding strength of the interface. A higher interfacial strength can minimize a localized debonding of the IPyC/SiC interface and avoid a stress concentration in the SiC layer, leading to a lower failure fraction of the particles [6].

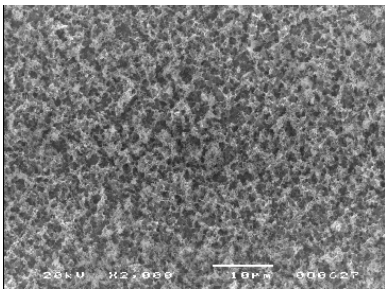

Fig. 3. A typical SEM microstructure for the inner surface of the SiC layer.

4. Conclusions

The fracture strength measured by crushing hemispherical shell specimens did not reveal a clear dependency on the coating temperature. This could be attributed to the large roughness of the inner surface of the SiC hemispherical shells in which the surface roughness acted as a fracture origin and overshadowed the porosity and microstructure effects in determining the fracture strength.

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