Sintering of Silicon Carbide and Theory of Sintering

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炭化ケイ素の焼結と焼結理論に関する研究

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The three important aspects of sintering of SiC, that is, powder processing, sintering additives and sintering theory, are reviewed in this paper. First, in order to obtain sinterable SiC powder, the carbothermic reduction of SiO₂ was investigated using a precursor derived from liquid sources. β (3C)-SiC powder free from any residual carbon was synthesized. The powder did not need a purification process and could be sintered with C and a small amount of B additives. Second, low-temperature sintering of α (6H)-SiC powder by AlB₂ and C additives was developed. AlB₂ addition enabled the sintering of SiC powder at around 1900°C. Al atoms from the additives dissolved into SiC and promoted elongation of grains during sintering. The fracture toughness of sintered SiC increased with the aspect ratio of the grains. Third, to understand the sintering behavior of SiC powder, new sintering models and a theory were proposed. In the theory, it was assumed that total excess energy in a system drove mass transport for sintering and grain growth. From the rate equations of sintering and grain growth, the surface energy and the grain boundary energy were found to play an important role in both processes.

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1. Introduction

SILICON carbide (SiC) ceramics are one of the most important engineering ceramics in the industry. They have good anticorrosive and superior tribological properties in addition to exhibiting high strength at high temperatures. They are widely applied in mechanical parts, such as mechanical seals, furnace substrates, high temperature parts and so on.^{1,2)} The high covalent bonding nature of SiC requires a very high temperature and suitable additives for sintering.¹⁾ The key points in the fabrication of high-density SiC materials are to obtain fine sinterable SiC powders and to find effective additives. Establishment of the basic theory on sintering is also important in the production of high-performance SiC materials.²⁾

In this paper, three key aspects of SiC materials are summarized, that is, the novel method of fabricating fine SiC powder,^{3,4)} the sintering process with the new additive^{5,6)} and the new sintering theory applicable to the processing of SiC materials.^{7,8)}

The powder characteristics are very important when fabricating sintered SiC. The most important characteristics are fine grain size, high purity and mono-phase crystal. In the industry, SiC powder is synthesized by the Acheson process or by carbothermic reduction of SiO₂ grains. The former process produces α (6H)-SiC crystals and SiC powder is obtained through pulverization and purification processes. The latter process produces β (3C)-SiC powder from SiO₂ and C powders. In this process, it is important to mix raw materials, because the carbothermic reduction of SiO₂ is a typical solid-state heterogeneous reaction. The synthesized powder requires a purification process to remove carbon residue.

This paper first describes the new process of SiC powder synthesis and sinterability of the powder. The author examined inorganic liquids as sources for SiO₂ and C to achieve fine mixing of raw materials. By polymerization and carbonization of these materials, a precursor consisting of SiO₂ and C was obtained and was converted into fine and sinterable

SiC powder. The powder was of pure β -SiC free from any residual carbon.

Second, the sintering of SiC powder using new additives of AlB₂ and C is described. Pressureless sintering of SiC by the addition of B and C was first reported by Prochazka.⁹⁾ This method has become an important technology in the manufacturing of SiC materials. There are, however, still some problems with this process, i.e., the high sintering temperature of $\sim$ 2150°C¹⁰⁾ and low toughness of the materials.^{11,12)} The author had intensively investigated the sintering process of SiC materials.^{13,14)} He found that AlB₂ and C were very effective additives.^{5,6)} These additives reduced the sintering temperature significantly and enhanced rod- or plate-shaped grain growth, which was expected to increase the fracture toughness of the materials.

Another important point in the sintering of SiC is to establish the sintering theory for the covalent bonding materials, such as SiC and Si₃N₄, which require additives for sintering.^{13,15)} The additives are thought to precipitate at the grain boundary,^{16,17)} promote sintering and control grain growth by decreasing grain boundary energy. One of the most important requirements for sintering SiC powder is to find suitable additives which can decrease grain boundary energy.¹⁾

Third, the new sintering and grain growth theories are described in this paper. There has been much work performed to date on the theory of sintering^{18)–24)} and grain growth.^{23,25)–28)} The traditional theories of sintering are based on the two contacting grains models. They assume that the difference in chemical potential due to the curvatures between the neck and grain surface drove mass transport. The theories can not explain directly the role of grain boundary energy on sintering and grain growth. A new sintering theory was required for the SiC materials. The new theory adopts the two-sphere models where two grains join at a planar grain boundary without neck curvature. The sintering proceeds by expansion of the grain boundary and grain growth proceeds by mass transport through the bound-

dary. Excess energy on the surface and at the grain boundary is assumed to drive mass transport. From the model, the rate equations for sintering and grain growth are derived, and the role of grain boundary energy on the sintering is discussed.

2. Synthesis of SiC powder and its sinterability

2.1 SiC precursor from liquid raw materials

Silicon-tetraethoxide ($((\text{C}_2\text{H}_5\text{O})_4\text{Si}(\text{OSi}(\text{OC}_2\text{H}_5)_2)_n)$) and phenolic resin were used for fine mixing of SiO_2 and C in a liquid phase. The mixtures of these materials with a suitable amount of water were stirred and polymerized. The polymers were dried and heated at 1000°C . Homogeneous glossy black SiC precursors were obtained. The precursors were glass-like solids, and SiO_2 and C components did not separate. The X-ray diffraction and IR analyses revealed that the precursors were amorphous and consisted of SiO_2 and C. The molar ratios of C/Si in the synthesized precursors were 1.7–3.1.

2.2 Synthesis of SiC powder

Fine yellow β -SiC powders were synthesized by the reaction of the precursors at 1500 – 1800°C . The X-ray diffraction patterns and the SEM image of the powders are shown in Figs. 1 and 2. The reaction of SiO_2 and C was completed at over 1600°C . The yields of SiC from the precursors were 0.6–0.8 depending on the C/Si ratio. For the C/Si ratios of 1.7–2.5, the powder contained neither residual SiO_2 nor C, and thus required no oxidation process or acid treatment. The particle sizes of the synthesized powders were in the sub-micron range for the synthesis temperature of 1600°C . The particle sizes increased to several microns for the synthesis temperatures of 1700 to 1800°C .

The X-ray diffraction analysis (Fig. 1) showed that the powder synthesized at 1600°C was of the β phase and contained a small amount of 2H phase. The powder synthesized at 1700°C was a pure β -SiC and had very large crystallite size (> 100 nm).

2.3 Sintering of SiC powder

The powder synthesized at 1750°C from the precursor

Fig. 1. X-ray diffraction patterns of the precursor and the synthesized powders.

C/Si=2.52 was pulverized and sintered by adding 0.1–1 mass% B and 2 mass% C. Figure 3 shows the results demonstrating high sinterability of the powder. The powder could be sintered nearly to the theoretical density with 0.2–1 mass% B and 2 mass% C. In the previous investigations of sintering of β -SiC powder, about 0.5–1 mass% B and 2 mass% were required to achieve a high density. Figure 4 shows the SEM micrograph of the sintered SiC. The microstructure was that of typical sintered β -SiC consisting of plate- and rod-shaped grains.

The powders from the precursors required no purification processes to remove either residual carbon or SiO_2 , and could be sintered by the addition of C and a small amount of B.

Fig. 2. Powder morphology synthesized at 1600°C from the precursor C/Si=2.52.

Fig. 3. Density of the SiC materials sintered from the powder synthesized from the precursor C/Si=2.52.

Fig. 4. Microstructure of the SiC material sintered at 2100°C with 0.2 mass% B and 2 mass% C. The starting powder was synthesized from the precursor C/Si=2.52 at 1750°C .

3. Sintering of SiC powder with AlB₂ and C additives

3.1 Low-temperature sintering of α -SiC powder by AlB₂ and C additives

α -SiC fine powder (mean grain size is 0.95 μ m), AlB₂ powder, and phenol resin (phenol formaldehyde condensate) for C were used as the starting material and additives. They were mixed in a SiC ball mill with ethanol medium. The contents of AlB₂ and C additives were 0.65–5.4 mass% and 2 mass%, respectively (Table 1). The powders were subsequently dried, compacted by CIP at 200 MPa, and sintered in a carbon resistance furnace. The specimens were first heated up to 1500°C in vacuum to remove SiO₂ in the SiC powder. They were then sintered between 1800 to 2150°C in Ar flow.

Table 1 lists the densities of the specimens sintered at various temperatures. The specimens which contained 0.67–2.7 mass% AlB₂ were sintered almost to the full density (>97% of TD (Theoretical density)) at a temperature as low as 1900°C. This sintering temperature is about 250°C lower than that of the conventional sintering process. Sintering of SiC powder in the industry is usually performed with B and C additives at about 2150°C. The specimen with 5.40 mass% AlB₂ did not sinter to the full density. Clearly exaggerated grain growth occurred at the early stage of sintering and suppressed densification of SiC powder.

Inomata and Tanaka²⁹⁾ investigated the phase compatibility in the Al–B–C system, and found the new compound Al₃B₄C₇, which formed a liquid phase at 1800°C. It was suggested that the liquid phase was formed from AlB₂ and C and enhanced sintering at 1800–1900°C due to a rapid diffusion path at the grain boundary.

3.2 Phase transformation of α -SiC and grain growth

The polytypes of SiC grains in the sintered materials were analyzed by the powder X-ray diffraction method. The results are listed in Table 2. The starting powder was mostly 6H-SiC, and part of 6H-SiC was transformed into 4H-SiC after sintering. The transformation was particularly marked when the sintering temperature was high and when the amount of AlB₂ was large. It was also observed that SiC grains grew aspherically in the specimens doped with a large amount of additives. Figure 5 shows sample of the microstructures of the sintered SiC.

Polymorphism and thermal stability of SiC have been studied in detail.^{1),2)} It has been clarified that 6H is stable at 2000°C and above. 4H is stable below 2000°C, and it becomes stable if Al atoms dissolve into SiC. Transforma-

tion observed in this study seems to be due to the Al solution from the additive AlB₂. 4H-SiC that contained Al atoms grew aspherically and was formed into rod- or plate-shaped grains during sintering.

3.3 Grain shape and mechanical properties

In general, grains in sintered materials have a round shape when they are made from 6H-SiC powder. In this experiment, however, SiC grains had an elongated shape. This grain shape was expected to improve the mechanical properties. Microstructure and fracture toughness were then measured with an image analyzer and by the indentation fracture method, respectively. The results are summarized in Table 3. The table shows that the fracture toughness of SiC materials increased in proportion to the grain size and the aspect ratio. Clearly, AlB₂ addition reduces the sintering temperature of SiC markedly, enhances aspherical grain growth, and thus contributes to the production of strong SiC materials.

Table 2. Polytype Contents in the Sintered SiC Specimens

Sample No.	Sintering Temp. (°C)	Polytype contents (%)		
		4H	6H	15R
6H-Starting powder		2	98	0
SC-1a	2000	4	92	4
SC-2a	2000	9	86	5
SC-3b	2150	30	67	3
SC-3c	1975	10	89	1
SC-3d	1975*	7	93	0
SC-4c	1975	4	85	11

Holding time at the maximum temperature is 30min or *0 min.

Table 1. Contents of Additives in Starting Powders and Sintering Densities

Sample	Contents of AlB ₂ , C and SiC powder (mass%)					
No.	AlB ₂	C		SiC powder		
SC-1	0.67	2.00		97.33		
SC-2	1.34	2.00		96.66		
SC-3	2.70	2.00		95.30		
SC-4	5.40	2.00		92.60		
Sintering Temperatures (°C) and densities of sintered SiC (%TD)						
	1800	1850	1900	1975	2000	2150
SC-1	87.2	94.6	98.0		98.8	99.9 ^G
SC-2	90.7	96.0	97.2		98.9	97.9 ^G
SC-3	90.2	96.5	97.4	97.8, 97.6*		96.7 ^G
SC-4	79.3	80.2	81.8 ^G	85.4 ^G		86.1 ^G

Holding time at the maximum temperature is 30min or *0 min.

^GThe sintered SiC materials exhibited large grain growth.

Fig. 5. Microstructures of the SiC material sintered from α (6H)-SiC powder with 2.7 mass% AlB₂ and 2 mass% C, (a) at 1975°C and (b) at 2150°C.

Table 3. Grain Size, Aspect Ratio and Fracture Toughness of Sintered SiC Material

Sample No.	Grain size (μm)	Aspect ratio	K_{IC} by IF method ($\text{MPa}\cdot\text{m}^{1/2}$)
SC-1a	1.7	1.6	3.6 ± 0.3
SC-2a	2.5	1.9	3.9 ± 0.2
SC-3b	6.1	2.6	4.3 ± 0.2
SC-3c	3.4	1.9	4.2 ± 0.3
SC-3d	3.1	1.8	4.1 ± 0.2
SC-4c	4.4	2.1	(porous)
SC-B*	1.3	1.7	3.0 ± 0.2

*The reference specimen sintered with 0.3mass% B and 2mass% C

4. New theory for sintering and grain growth

4.1 Models for sintering and grain growth

When SiC powder compacts are heated at high temperatures, grains of SiC begin to sinter. During sintering, the surface area of the grains converts to the grain boundary area, and the total surface and grain boundary energy decreases. This decrease in energy is the driving force for sintering. The new sintering and grain growth theories developed from the total free energy theory^(7,8) are described in this section.

Two-sphere grain models for sintering and grain growth are proposed as shown in Fig. 6. In the models, two spherical grains of different size are in contact forming grain boundary. Sintering is defined as a process of reducing the distance between two grain centers and expanding grain boundary area with constant grain volumes, and grain growth is defined as a process where material is transported from a small grain to a large grain with a constant grain boundary area (Fig. 6).

The following non-dimensional parameters are defined for the calculation.

$$R_0 = \frac{r_{02}}{r_{01}}, R_1 = \frac{r_1}{r_{01}}, R_2 = \frac{r_2}{r_{02}},$$

$$X_1 = \frac{x_1}{r_1}, X_2 = \frac{x_2}{r_2}, \alpha = \frac{\epsilon_{gb}}{\epsilon_s} \quad (1)$$

where r_{01} and r_{02} are the initial radii of grains 1 and 2, r_1, r_2 and x_1, x_2 are grain radii and the distances between the grain center and the boundary of grains 1 and 2, and ϵ_{gb} and ϵ_s are the grain boundary energy and the surface energy, respectively.

The grain size ratio is $0 < R_0 \leq 1$ and the ratio of the surface energy and the grain boundary energy is $0 \leq \alpha \leq 1$. During sintering, the distance of grains X_1 and X_2 decreases with expanding grain boundary. During grain growth, R_0 decreases by mass transport from the smaller grain to the larger grain.

4.2 Driving energy for mass transport

The basic concept of the theory is that mass transport for sintering and grain growth in the system is proportional to the diffusion coefficient, D_x , the term of diffusion path (a_x/λ_x) and the excess free energy, ΔG , in the system. The rate of mass transport, dv/dt , and (a_x/λ_x), therefore, are given by the following equations.^(30,31)

$$\frac{dv}{dt} = D_x \left(\frac{a_x}{\lambda_x} \right) \left\{ 1 - \exp \left(\frac{-\Delta G}{RT} \right) \right\} \quad (2)$$

Fig. 6. Two-sphere models (a) for sintering and (b) for grain growth.

$$\left(\frac{a_x}{\lambda_x} \right)^{-1} = \int_{\text{sys}} \left(\frac{a_m}{\lambda_x} \right)^{-1} dv \int_{\text{sys}} dv$$

$$a_m = (a_i a_0)^{1/2} \quad (3)$$

Where, (a_x/λ_x) is given by the harmonic average of the mean diffusion area, a_m , divided by the diffusion length, λ_x , over the system, where a_m is defined as the mean cross-sectional area of mass flow from the area a_i to the outlet a_0 . R and T are the gas constant and temperature, respectively.

4.3 Rate of grain shrinkage due to sintering

Referring to Fig. 6, the sum of surface and grain boundary energies, Ψ , which is normalized by the initial surface energy, is:

$$\Psi = \frac{\{\epsilon_s(s_1 + s_2) + 2\epsilon_{gb}s_{gb}\}}{4\pi\epsilon_s(r_{01}^2 + r_{02}^2)}$$

$$= \frac{R_1^2(1 + X_1) + R_0^2 R_2^2(1 + X_2) + \alpha(1 - X_1^2)}{2(1 + R_0^2)} \quad (4)$$

Ψ decreases as X_1 and X_2 decrease until Ψ reaches a minimum, $\Psi_{\min}$. The system is in an equilibrium state at $\Psi_{\min}$. The system in the present state has the excess energy, $\Delta\Psi = \Psi - \Psi_{\min}$, which drives mass transport for sintering. ΔG_s for the sintering in Eq. (2) for one mole is expressed as:

$$\Delta G_s = \frac{3\epsilon V_m}{r_{01}} \Delta\Psi(1 + R_0^2)/(1 + R_0^3), \quad (5)$$

where V_m is the mole volume.

The term of the diffusion path, (a_x/λ_x), is obtained by integration of Eq. (3) with respect to the whole volume of the two-sphere model (Fig. 6). The calculation gives the following equations.⁽⁸⁾

$$\left(\frac{a_x}{\lambda_x} \right) = 16\pi r_{01} g_d$$

$$g_d = (1 + R_0^3) \left[\frac{R_1^2}{(1 - X_1^2)^{1/2}} \{4(1 - X_1^2)^{3/2} + 6X_1^2(1 - X_1^2)^{1/2} + 6X_1 \arcsin X_1 + 3\pi X_1\} \right.$$

$$\left. + \frac{R_0^2 R_2^2}{(1 - X_2^2)^{1/2}} \{4(1 - X_2^2)^{3/2} + 6X_2^2(1 - X_2^2)^{1/2} + 6X_2 \arcsin X_2 + 3\pi X_2\} \right]$$

$$+ 6X_2 \arcsin X_2 + 3\pi X_2 \left]^{-1} \quad (6)$$

It is also necessary to calculate the amount of mass required for sintering or for reducing X_1 and X_2 . This is $dv = (dX_1 + dX_2) \times (\text{the grain boundary area})$, and is expressed as:

$$\frac{dv}{dt} = - \frac{2\pi r_{01}^3 R_1^2 \{R_1(1-X_1) + R_0 R_2(1-X_2)\}}{(2-X_1)} \frac{dX_1}{dt} \quad (7)$$

From Eqs.(2),(4),(5)–(7), the shrinkage rate, dX_1/dt , for grain 1 due to sintering is calculated as follows.

$$\begin{aligned} \frac{dX_1}{dt} = & - \left(\frac{3\epsilon_s D_x V_m}{\pi RT} \right) \frac{8\pi}{r_{01}^3} \\ & \times \frac{(1+R_0^2)(2-X_1)\Delta\Psi g_d}{(1+R_0^3)R_1^2\{R_1(1-X_1) + R_0 R_2(1-X_2)\}} \end{aligned} \quad (8)$$

dX_2/dt for grain 2 is obtained by conversions of X_1 to X_2 and X_2 to X_1 , R_1 to R_2 and R_2 to R_1 , and R_0 to $1/R_0$ in Eq.(8).

4.4 Growth rate of grains

The growth or shrinkage of two grains is driven by the excess surface energy. The excess energies, ϕ_1 and ϕ_2 , for the grain growth of two grains 1 and 2, respectively, are estimated by the difference in surface energies between grains 1 and 2 and the system. ϕ_1 and ϕ_2 , and the excess energy, ΔG_g , for the system are the mole average of $\phi_1(>0)$ and $-\phi_2(>0)$ and are given by the following equations,^{7,8)}

$$\begin{aligned} \phi_1 &= \frac{\{\epsilon_s(s_1+s_2) + 2\epsilon_{gb}s_{gb}\}}{4\pi(r_{01}^3+r_{02}^3)} - \frac{(\epsilon_s s_1 + \epsilon_{gb}s_{gb})}{4\pi r_{01}^3} \\ \phi_2 &= \frac{\{\epsilon_s(s_1+s_2) + 2\epsilon_{gb}s_{gb}\}}{4\pi(r_{01}^3+r_{02}^3)} - \frac{(\epsilon_s s_2 + \epsilon_{gb}s_{gb})}{4\pi r_{02}^3} \\ \Delta G_g &= \frac{r_{01}^3}{r_{01}^3+r_{02}^3} \phi_1 - \frac{r_{02}^3}{r_{01}^3+r_{02}^3} \phi_2 = \frac{3\epsilon_s V_m}{r_{01}} f_g \\ f_g &= - \frac{1}{(1+R_0^3)} \left[- \left\{ \frac{1-R_0^3}{2(1+R_0^3)} \right\} \{R_1^2(1+X_1) \right. \\ & \quad + R_0^2 R_2^2(1+X_2) + \alpha R_1^2(1-X_1^2)\} \\ & \quad + \frac{R_1^2}{4} \{2(1+X_1) + \alpha(1-X_1^2)\} \\ & \quad \left. - \frac{R_0^2 R_2^2}{4} \{2(1+X_2) + \alpha(1-X_2^2)\} \right] \end{aligned} \quad (9)$$

where ϕ_1 is positive and grain 1 grows, and ϕ_2 is negative and the smaller grain 2 shrinks. The total mass required for grain growth is given as:

$$\begin{aligned} \frac{dv}{dt} &= 2 \times 4\pi r_{01}^2 \frac{dr_{01}}{dt} \\ 4\pi r_{01}^2 \frac{dr_{01}}{dt} &= -4\pi r_{02}^2 \frac{dr_{02}}{dt} \end{aligned} \quad (10)$$

From Eqs.(2), (6), (9) and (10), the growth rate for grain 1 is calculated using Eq.(11).

$$\frac{1}{r_{01}} \frac{dr_{01}}{dt} = \left(\frac{3\epsilon_s D_x V_m}{\pi RT} \right) \frac{2\pi}{r_{01}^3} f_g g_d \quad (11)$$

The growth rate for grain 2 is obtained using Eqs.(10) and

(11).

4.5 Calculations of sintering shrinkage and grain growth

Sintering and grain growth are assumed to take place simultaneously. The sintering shrinkage, dX_1 and dX_2 , and the changes of grain radii, dr_{01} and dr_{02} , were calculated from Eqs.(8) and (11) for the infinitesimal time, dt . They were sequentially summed up. The calculation began at X_1 and $X_2=1$ (time=0), and continued until X_2 reached 0. Some of the results are shown in Fig. 7. The sintering behaviors of two grains in Fig. 7 are represented as a function of α in Fig. 8. In Fig. 7 the changes in X_1 and r_{02} are plotted under the conditions, for example, of constant α (=0.5) with varying R_0 or constant R_0 (=0.5) with varying α .

4.6 Sintering and grain growth behavior of two grains

The calculations of sintering behavior (Figs. 7 and 8) clearly revealed the effects of the grain boundary energy. When the ratio of the grain boundary to surface energy, α ,

Fig. 7. Changes of X_1 and r_{02} with time Bt . $B (= 3\epsilon_s D_x V_m / \pi RT)$ is a normalizing factor. The initial grain size ratio, $R(0)$, was held constant at 0.5 while the ratio of grain boundary energy to surface energy, α , was varied from 0 to 0.98 in (a) and (c). α was held constant at 0.5 while $R_0(0)$ was varied from 0.1 to 1 in (b) and (d).

Fig. 8. Schematics of sintering behavior of two joining grains. The sintering (grain boundary expansion) rate became faster and the grain shrinkage rate of the smaller grain became slower as α increased from 0 to 0.9.

was small, X_1 rapidly reached 0, while r_{02} did not decrease significantly. Two grains coalesced in a short period of time, but grain growth was not fast. This means that a powder can be sintered very easily when its grain boundary energy is small. When α was nearly 1, X_1 did not change remarkably, and almost retained its initial value of 1. This means that a powder with high grain boundary energy can sinter negligibly. There was also a tendency for grain growth to take place dominantly when the size ratio of two grains (R_0) was small. In this case, the grain boundary did not expand markedly and sintering proceeded slowly.

The calculations performed here demonstrate that the smaller the grain boundary energy and the smaller the difference in grain size, the faster the sintering occurs. It is therefore suggested that the pure SiC powder hardly sinters, because SiC has a highly covalent bonding nature and the grain boundary energy is almost as large as the surface energy.^{32,33} The additives, such as B and C, or AlB_2 and C, are necessary for decreasing the grain boundary energy. Fine SiC powders with uniform grain distribution and suitable additives sinter readily, as shown in the previous sections 2 and 3.

5. Summary

In this paper, the three important aspects of SiC materials, that is, the powder synthesis, the sintering process, and the theory of sintering and grain growth are reviewed. They are summarized as follows.

(1) SiC precursors were derived from the liquid materials of silicon-tetraethoxide and phenolic resin. Homogeneous mixing of SiO_2 and C could be achieved in the precursor. The pure SiC powders without residual carbon were obtained by carbothermic reduction of the precursors. SiC powders were of the single-phase $\beta(3\text{C})$ type, and sintered with the addition of carbon and a small amount of B.

(2) The low-temperature sintering of SiC powder was developed using AlB_2 and C additives. These additives could sinter SiC powder at around 1900°C , and promoted elongation of α -SiC grains during sintering. Al atoms from the additive appear to have dissolved into SiC and partly transformed 6H-SiC into 4H-SiC . This transformation caused anisotropic grain growth. The fracture toughness of the sintered SiC materials increased as the aspect ratio and the size of grains increased.

(3) New models and theory for sintering and grain

growth were proposed. The theory assumed that the total free energy on the surface and grain boundary of grains drove mass transport. The rate equations for sintering and grain growth were derived from the theory, and the sintering behavior involved in joining two grains was simulated. It is found that the grain boundary energy, which was dependent on the additives used, was important for sintering of SiC, and that low grain boundary energy was necessary for stimulation of sintering and suppression of grain growth.

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