

Effect of carbon source on the properties of dense \(\alpha\)-SiC

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Abstract

Due to its outstanding properties, SiC is a candidate material for use in special applications such as armor. In order to use SiC in these special applications, it is necessary to produce fully dense ceramics. The ability to produce high density materials with superior performance depends on a number of factors. One of these factors is the addition of carbon to aid sintering. In this study, the effect of different carbon sources and ratios on the elastic and mechanical properties of SiC was investigated. Two types of carbon (lamp black and phenolic resin) were added to SiC in different ratios (0\%–2\% wt.). All samples were sintered via the spark plasma sintering (SPS) method at 1900 °C for 15 min and under 50 MPa pressure. Samples made with lamp black were reached full density at 1.0 wt.%C, and the hardness and elastic modulus values were \(\sim\)22 GPa and 440 GPa, respectively. While samples made with both carbon sources showed similar bulk mechanical properties, the samples made with lamp black showed more consistent microstructures. The carbon from the phenolic resin source did not appear to be as well distributed as that from the lamp black source. The results also confirmed that addition of carbon into SiC was essential to improve the density and other mechanical properties associated with it.

1. Introduction

Silicon carbide (SiC) is a good candidate to use in industrial applications due to its exceptional properties such as low density (3.21 g/cm\(^3\)), low coefficient of thermal expansion, high elastic modulus, hardness, and thermal conductivity, and good wear and oxidation resistance [1–9]. Therefore it is widely used in applications such as armor, abrasives, rocket nozzles, high temperature, corrosion, and wear resistant applications, and aerospace [10–15].

One challenge with silicon carbide is that sintering is difficult due to its strong covalent bonding and low self-diffusion coefficient [16, 17]. In order to eliminate this problem, either a sintering aid (B\(_2\)O\(_3\), B, Al, Si, TiC, ZrB\(_2\), TiB\(_2\), Al\(_2\)O\(_3\), Y\(_2\)O\(_3\), and C) must be used or it must be sintered at high pressure (hot press, spark plasma sintering) and/or high temperature [18–28]. However using oxide additives can cause the formation of a liquid phase at the sintering temperature which can decreases the mechanical properties of SiC due to the residual glassy phase [29, 30].

Another important factor affecting the sintering of non-oxide ceramic materials like SiC is the native oxide layer on their powder surface. For silicon carbide, this layer occurs as silicon dioxide (SiO\(_2\)). The presence of this oxide layer can contribute to coarsening and inhibit densification during sintering [31–33]. The addition of boron and carbon as sintering aids seems to be the best candidate to take advantage of the good properties of SiC, as it aids in solid state sintering and does not form a liquid phase. Boron containing additives increase the densification rate of SiC by increasing the self diffusion of the system [34]. On the other hand, carbon eliminates the oxide layer on the SiC surfaces preventing excessive grain growth [20, 21, 23, 25, 28]. Stobierski et al showed
that both boron and carbon should be added together to SiC in order to produce highly dense parts since glassy phase formation was observed in samples with only boron added \cite{25, 34}.

The goal of this study was to evaluate the impact of the carbon source and content on the microstructure, hardness, and elastic moduli of dense silicon carbide materials. In order to achieve this objective, silicone carbide powders with additions of 0.5 wt.% boron carbide and 0–2 wt.% carbon added as either lamp black or phenolic resin, were densified by spark plasma sintering. The dense silicone carbide ceramics were then characterized using the Archimedes method, FESEM with EBSD, ultrasound, and Knoop hardness techniques to determine the effects of the different carbon sources on the microstructure and mechanical properties.

2. Experimental

2.1. Powder preparation

Saint Gobain (Niagara Falls, New York) α- silicon carbide powder, synthesized by the Acheson process and having 1.5 μm average particle size, was used in this study. H.C. Starck boron carbide (HD-20, H. C. Starck GmbH&Co., Germany), carbon lamp black (Fisher Scientific, USA) and phenolic resin (VARCUM 29353, Durez Corp.) were used as additives.

For the samples made with lamp black as the carbon source, called the SG-LBC series, SiC, 0.5 wt.% B₄C and 0–2 wt.% lamp black were mixed by ball milling in ethanol with SiC media for 24 h in a HDPE Nalgene bottle. The slurry was then sieved to remove the SiC media and pan dried. For each sample a 50 gram batch of powder was prepared. The SG-LBC series compositions are shown in table 1.

The amount of phenolic resin needed to be able to generate the same amount of carbon as the SG-LBC series samples (0.5%, 1.0%, and 1.5%) was calculated from the char yield. Thermogravimetric analysis (TGA) was performed on the resin, and showed that the char yield was approximately 43% as shown in figure 1. According to the TGA result, the weight of the added phenolic resin should be 2.58 times the amount of carbon desired in the final formulation. The compositions of the samples made with the phenolic resin carbon source, called the SG-PRC series, are shown in table 2.

The liquid phenolic resin was mixed with deionized water and sonicated to help disperse it. After that SiC and B₄C powders, ammonium hydroxide (0.22 g) and SiC milling media were added to the liquid mixture and ball milled for 24 h. The slurry was then sieved to separate SiC media from liquid mixture. To remove the excess water, the liquid mixture was filter pressed at 35 psi and dried in an oven at 100 °C.

![Figure 1. VARCUM phenolic resin TGA data.](image-url)
2.2. Sintering of samples

6.5 grams of dry powder mixture were put in a graphite die (20 mm inner diameter) and densified using a Thermal Technology SPS 10-4 spark plasma sintering unit (Thermal Technology, LLC, Santa Rosa, CA, USA). To prevent possible reaction between powder and die, the die and punches were lined with graphite foil.

To densify the SG-LBC series samples, the powder was heated to 1400 °C at a 200 °C per minute heating rate under vacuum with applied pressure of 50MPa and held under those conditions for 30 min. After the 30 min holding time at 1400 °C, the SPS chamber was backfilled with argon and heated to 1900 °C at a 200 °C per minute heating rate, maintaining 50 MPa applied pressure and held for 15 min at 1900 °C. After sintering, the SPS powder was turned off and the sample was allowed to cool.

To densify the SG-PRC series samples, an additional step was added to convert the resin into carbon. The SPS was first heated to 800 °C at a 200 °C per minute heating rate under vacuum with applied pressure of 20MPa and held for one hour. Afterwards, the same sintering conditions as the SG-LBC samples were applied. The SPS conditions can be seen in figure 2.

2.3. Post sintering processing

After densification, the samples were covered with excess graphite foil. This graphite foil was removed by sand blasting using garnet blasting media. Since the samples needed to be smooth for ultrasound analysis, the samples were ground flat using a surface grinder with a 600 grit diamond wheel. The samples were then cut into small pieces using a LEKO VC-50 diamond saw (LECO Corporation, St. Joseph, MI, USA), mounted in epoxy resin using a Buehler SimpliMet 1000 automatic mounting press (Buehler, Lake Bluff, IL, USA), and polished to 0.25 μm finish using a Buehler EcoMet 250 automatic grinder/polisher with AutoMet 250 automatic head in preparation for FESM analysis and microhardness testing. Polished samples were also ion milled using the flat-milling mode with 3 kV acceleration voltage, 80° tilt angle, no offset, and 25 rpm rotation speed for 10 min using a Hitachi IM4000 (Hitachi High-Technologies Corporation, Tokyo, Japan) to prepare for EBSD analysis. Samples were etched using a modified Murakami method (20 g KOH and 20 g K$_3$Fe(CN)$_6$ in 60 ml deionized water) in order to highlight grains boundaries to better view the microstructure with FESEM.
2.4. Dense sample characterization
All sample densities were measured by the Archimedes method after surface grinding. Elastic properties were measured using ultrasound analysis. Hardness of the samples were measured using a Knoop diamond indenter by making 10 indents each at loads of 100, 200, 500, 1000, and 2000 grams with a LECO microhardness tester. Sizes of the indents were measured using a Keyence VHX5000 digital microscope (Keyence Corporation, Osaka, Japan). To examine the microstructure of the samples, Zeiss Sigma field emission scanning electron microscope (FESEM) was used. The grain size of each sample was measured (minimum of 100 intercepts were measured) by the linear intercepts method using Lince 2.42 software. The electron backscatter diffraction (EBSD) map of each sample was obtained using an Oxford Instruments Nordlys Nano EBSD detector (Oxford Instruments, Abingdon, UK) on the Zeiss Sigma FESEM.

3. Results and discussion
3.1. Microstructure characterization
Figure 3 shows the microstructures of the SG-LBC series of samples. The grain morphology of all samples were similar and showed equiaxed grain shapes. Only their grain size and their porosity differ according to the amount of carbon they contain. The sample made without carbon addition showed significant amounts of porosity. Addition of 0.5 wt.% C helped reduce the porosity, but was not enough to reach full density. It can be seen from the microstructure images that, by adding 1.0 wt.% and 1.5wt.% C, completely dense SiC samples
were obtained. Pores were not observed in the structure. However, some porosity was detected again in the sample with 2.0 wt.% C. Due to the native oxide layer of SiC, it is difficult to obtain high density SiC without carbon addition. To eliminate the oxide layer, it was essential to add sufficient amounts of carbon [8, 17]. It can be seen from FESEM images that increasing the amount of carbon from 0 to 1.0 wt.%, decreased the porosity and the SiC ceramic reached the full density at this point (1.0 wt.%C). However, the grain size of the samples increased with the added carbon ratio from 4.08 μm to 5.43 μm, since there were a few pores to prevent the grain growth. With increasing additions of carbon, samples maintained high density but the average grain size of samples was reduced from 5.43 μm to 2.66 μm due to grain boundary pinning by inclusions of the excess carbon. Similar effects could be seen in the literature [25]. The change of average particle size and standard deviation with carbon content is shown in table 3.

The microstructures of the SG-PRC series of samples can be seen in figure 4. The sample made without C, with 0.5 and 1.0 wt.% C also showed significant amounts of porosity. The sample made with addition of 1.5wt.% C is almost fully dense (>99%), and had very few small pores. The grain morphology and the average grain size of these samples were quite similar. The density values and average grain sizes with standard deviation can be seen in table 4. The primary difference between these four samples was the amount of porosity. Adding 1.0 wt.% C enabled to achieve high density for the LBC series. The same result was expected in the PRC series, but this did not happen. When carbon is added to the SiC, the C reacts with the oxide on the surface of the SiC at high temperature and forms volatile oxygen containing species, which are removed via gas transport. In cases where a small amount of added C is not sufficient to reduce the oxygen content, it is prevented densification. In such a

| Table 3. SG-LBC series elastic and microstructural properties. |
|---------------------------------------------------------------|
| Sample          | Density (g/cm³) | Average grain size (std.dev.) | Phase fraction 4H (%) | Phase fraction 6H (%) | 4H/6H ratio | E (GPa) |
|-----------------|----------------|-----------------------------|---------------------|---------------------|-------------|---------|
| SG-0 C          | 3.08           | 4.08 (0.79)                 | 10                  | 69                  | 0.14        | 424     |
| SG-LBC-0.5 C    | 3.09           | 3.13 (0.33)                 | 11                  | 69                  | 0.16        | 416     |
| SG-LBC-1.0 C    | 3.19           | 5.43 (0.89)                 | 17                  | 71                  | 0.24        | 440     |
| SG-LBC-1.5 C    | 3.18           | 4.92 (0.30)                 | 10                  | 73                  | 0.14        | 431     |
| SG-LBC-2.0 C    | 3.13           | 2.66 (0.57)                 | 11                  | 62                  | 0.18        | 411     |

Figure 4. Microstructure of SG-PRC series.
| Sample         | Density (g/cm³) | Average grain size (std.dev.) | Phase fraction 4H (%) | Phase fraction 6H (%) | 4H/6H ratio | E (GPa) |
|----------------|----------------|------------------------------|-----------------------|-----------------------|-------------|---------|
| SG-PRC-0.5 C   | 3.05           | 3.08 (0.76)                  | 10                    | 68                    | 0.15        | 407     |
| SG-PRC-1.0 C   | 3.08           | 3.92 (0.56)                  | 10                    | 66                    | 0.23        | 415     |
| SG-PRC-1.5 C   | 3.18           | 3.25 (0.58)                  | 10                    | 57                    | 0.18        | 441     |
case, it is expected that no carbon will remain in the microstructure as all of the carbon will react with oxygen. However, this was not the case for the SG-PRC series, as there were visible carbon residues in the microstructures of the samples. It can be thought that the carbon from this source accumulates in some regions due to inhomogeneous mixing, and reduces the oxygen content locally, but after sintering, leaves residual carbon in those regions. In areas where there was less carbon, the oxide is not completely removed and results in residual porosity. When comparing both series, the SG-PRC series showed higher amounts of porosity than SG-LBC series when the same amounts of carbon was added. This suggested that the phenolic resin may not mixed as well as the lamp black or that more of the phenolic resin might be lost during sintering than predicted by the TGA, so the remaining carbon was not enough to fully remove the surface oxide layer.

EBSD maps of the SG-LBC series of samples can be seen in figure 5. The red color in the maps indicates the 6H-SiC polytype and the green color the 4H-SiC polytype. Black color indicates areas where no SiC phase could be detected because of the presence of secondary phases, pores, roughness, grain boundaries or other factors. It can be seen from the EBSD maps that the samples had a predominantly 6H polytype, and a small proportion of the 4H polytype. The phase fraction of 6H and 4H of SG-LBC series can be seen in table 3. As the phase fractions were determined by the areas of each color in the image, the sum of the 6H and 4H polytypes were not 100% due to the presence of the black areas. The raw SiC powder’s 4H/6H ratio was 0.11, and the ratio of 4H/6H for the SG-LBC series were between 0.14–0.24 which were slightly higher than the starting powder ratio. It can be concluded that addition of lamp black carbon had effect on the conversion of 6H polytype to 4H polytype. This inference is also supported by the literature [5].

Figure 6 shows EBSD maps of the SG-PRC series of samples. The EBSD maps of SG-PRC samples are similar to those of the SG-LBC samples. The phase fractions of 6H and 4H for the SG-PRC series can be seen in table 4.
A similar situation to the SG-LBC series is observed in these samples. Samples mostly contain the 6H-SiC polytypes with smaller amounts of 4H-SiC polytype with the ratio of 4H/6H between 0.15–0.23. As seen in the EBSD maps and FESEM images, all samples showed primarily equiaxed grains of mostly 6H-SiC polytype. This result is supported by previous research the literature shows that 6H-SiC polytype smaller and equiaxed grains and 4H-SiC polytype shows large and elongated grains [8].

3.2. Elastic and mechanical properties characterization

The densities and elastic modulus values of the SG-LBC series samples are shown in table 3. When looking at sample densities, it can be seen that 0.5 wt.% C was insufficient to remove the oxide layer and achieve densification, it only reached theoretical density of 96%. The highest density (>99%) was obtained at 1.0 wt.% C addition. This was sufficient to remove the oxide layer from the SiC surface. When 1.5 wt.% C was added, a slightly decrease in density was seen. However, since the high density was already reached at 1.0 wt.% C, the 1.5 wt.% C was more than the carbon required to remove oxygen from the SiC surface, and this excess carbon caused a decrease in the average grain size. At the highest C levels, there was excess carbon residues in the microstructure. Because the theoretical density of carbon is lower than that of SiC, residual carbon in the structure caused the density of the sample to decrease. Lomello et al sintered SiC without additives by SPS at 1900 °C for 5 min and under 70 MPa pressure, but due to the high oxygen content of the powder and not adding carbon to remove it, only 96% density was achieved [19].

Similar effects can be seen in the elastic modulus of the samples. The highest elastic modulus was achieved at 1.0 wt.% C since it had the highest density. The elastic modulus of materials has close relationship with materials...
density and secondary phase inclusions. It can be seen from the table that lower elastic modulus were obtained at low carbon additions because of porosity, while the modulus values of samples with larger carbon additions were reduced by the residual carbon. The 0 wt.% C and 0.5 wt.% C are very similar in density and microstructure and that the difference in elastic modulus is probably due to variability in the ultrasound measurement. The standard deviation value of this elastic modulus is ±5. A similar effect can be seen with conventional sintering method, since SiC sintered without applied pressure could not reach high density and had pores in the microstructure and could only reach the elastic modulus of 409 GPa [11].

The density values of the SG-PRC series samples and elastic modulus values are shown below in table 4. It can be seen from the results that increasing amounts of added carbon results in an increased in density. While only 95% of the theoretical density was achieved when 0.5 wt.% C was added, a theoretical density of >99% could be reached when 1.5 wt.% C was added.

Looking at the work of Guillard et al, it is clearly seen how important the additives are in the sintering of SiC. SiC samples sintered using SPS at 75 MPa pressure, 5 min at a temperature that could be considered as high as 1850 °C, only 92% density was achieved [30].

Load-Knoop hardness curves for SG-LBC series samples are shown below in figure 7. It can be clearly seen in the graph that the hardness of the samples decreased with the increasing loads. This behavior is seen in ceramic materials in general [35]. The effect of carbon addition on the hardness of SiC also follows the similar trend to the
effect on the elastic modulus. Hardness values increased with the density of the SiC, and obtained the highest value at 1.0 wt.% C. As increasing the carbon addition from 0.5 to 1.0 wt.%, first hardness value increased from 19.89 to 21.98 GPa at 1000 g load. Then increased the carbon content from 1.0 to 1.5 wt.%, the hardness values slightly decreased from 21.98 to 21.34 GPa at 1000 g load. Increasing the carbon addition further, at 2.0 wt.% C, the hardness value was 20.83 GPa at 1000 g load. At 0.5 and 2.0 wt.% C the low hardness values were due to the effects of porosity and residual carbon. This effect of porosity has also been seen in other studies [19]. It can be concluded that just enough carbon should be added to SiC to produce dense samples with better elastic and mechanical properties, as residual carbon also has a negative effect on SiC mechanical properties.

Load-knoop hardness curves for SG-PRC series samples are shown below in figure 8. By increasing the amount of added of carbon from 0.5 to 1.5 wt.%, the hardness values increased from 19.35 to 21.43 GPa at 1000 g load. This is again because of the increasing densities of samples with addition of carbon. The highest value was obtained at 1.5 wt.% C. It has been seen in previous studies that with the addition of more than 1.5–2 wt.% carbon to SiC, residual carbon is easily seen in the microstructure and reduced the hardness of SiC [17, 36]. Therefore, in this study, phenolic resin was added to SiC at a maximum of 1.5wt.%. This was slightly different from the SG-LBC series where the highest hardness was obtained at 1% C before the hardness decreased as more carbon was added. This may be because more of the carbon from the phenolic resin was lost during processing than anticipated or because the phenolic resin was not homogeneously distributed so that it could effectively remove all of the surface oxide.

4. Conclusions

To investigate the effect of different carbon additives on the properties of SiC, two different carbon sources were added to SiC with 0.50 wt.% B₄C. The carbon sources were particulate lamp black (LBC), and liquid phenolic resin (PRC). All samples were densified by SPS at 1900 °C for 15 min under 50 MPa applied pressure in argon atmosphere. This sintering temperature and time were sufficient for high density SiC production.

The general results can be listed as follows;

- Samples made with low carbon content or without carbon showed significant amounts of porosity.
- Low elastic, and mechanical properties were obtained at low carbon addition because of porosity, or with high amount of carbon addition due to residual carbon.
- Addition of either type of carbon had affect the conversion of 6H to 4H SiC polytype.
- Addition of carbon of both types affected the density, average grain size, hardness, and elastic properties.

For the SG-LBC series, the highest hardness, elastic modulus, and density were obtained at 1.0 wt.% C addition, and the values were ~22 GPa, 440 GPa, and >99%, respectively. For the SG-PRC series the highest hardness, elastic modulus, and density were obtained at 1.5 wt.% C addition, and the values were 21.43 GPa, 441 GPa, and >99%, respectively. Although the results were similar for both series, differences in microstructural properties were observed. The results suggest that the phenolic resin was not able to be mixed as well as the lamp black carbon. These results showed that uniform addition of carbon to SiC is essential for obtaining high density and other properties associated with it. However, only enough carbon must be added to the SiC to remove the surface oxide layer as the presence of residual carbon can adversely affects the mechanical properties of SiC.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).
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References

[1] Liu M, Yang Y, Wei Y, Li Y, Zhang H, Liu X and Huang Z 2019 Preparation of dense and high-purity SiC ceramics by pressureless solid-state-sintering Ceram. Int. 45 19771–6
[2] Sivkov A, Nikitin D, Shanenko I, Ivashutenko A, Rahmatullin I and Nasyrybayev A 2019 Optimization of phase dynamic synthesis of ultradisperse silicon carbide and obtaining SPL ceramics on its basis Int. J. Refract. Met. Hard Mater 79 123–30
[3] Cho T-Y and Kim Y-W 2017 Effect of grain growth on the thermal conductivity of liquid-phase sintered silicon carbide ceramics J. Eur. Ceram. Soc. 37 3475–81
[4] Khodaei M, Yaghobizadeh O, Baharvandi H R and Dashki A 2018 Effects of different sintering methods and conditions on the properties of SiC-TiC, SiC-TiB2 composites Int. J. Refract. Met. Hard Mater 70 19–31
[5] Kang K-H, Eun T, Jun M-C and Lee B-J 2014 Governing factors for the formation of 4H or 6H-SiC polytype during SiC crystal growth: an atomistic computational approach J. Cryst. Growth 389 120–33
[6] Alhosseini S H N and Mouavi S R 2019 The effect of oxide, carbide, nitride and boride additives on properties of pressureless sintered SiC A review J. Eur. Ceram. Soc. 39 2215–31
[7] Yasar Z A and Haber R 2020 Effect of acid etching time and concentration on oxygen content of powder on the microstructure and elastic properties of silicon carbide densified by SPS International Journal of Materials Science and Applications 9 7
[8] Yasar Z A, DeLuca V A and Haber R A 2018 Influence of oxygen content on the microstructure and mechanical properties of SPS SiC Ceram. Int. 44 23248–53
[9] Tanaka H, Yoshimura H N, Otani S, Zhou Y and Toriyama M 2000 Influence of silica and aluminum contents on sintering of and grain growth in 6H-SiC powders J. Am. Ceram. Soc. 83 226–8
[10] Zhang J, Jiang D, Lin Q, Chen Z and Huang Z 2015 Properties of silicon carbide ceramics from gelcasting and pressureless sintering Materials & Design (1980–2015) 65 12–6
[11] Magnani G, Brentari A, Burresi E and Raiteri G 2014 Pressureless sintered silicon carbide with enhanced mechanical properties obtained by the two-step sintering method Ceram. Int. 40 1759–63
[12] Liang H, Yao X, Huang Z, Zeng Y and Su B 2016 Effect of sintering techniques on the microstructure of liquid-phase-sintered SiC ceramics J. Eur. Ceram. Soc. 36 1863–71
[13] Saddow S E and Agarwal A K 2004 Advances in silicon carbide processing and applications Acta Tech. House 1–67
[14] Moskovskikh D, Song Y, Rouvimov S, Rogachev A and Mukasyan A 2016 Silicon carbide ceramics: mechanical activation, combustion and spark plasma sintering Ceram. Int. 42 12686–93
[15] Madar R 2004 Silicon carbide in contention Nature 430 974–5
[16] Kim Y W, Mitomo M, Emoto H and Lee J G 1998 Effect of initial α-phase content on microstructure and mechanical properties of sintered silicon carbide J. Am. Ceram. Soc. 81 3136–40
[17] Yasar Z A and Haber R A 2020 Effect of carbon addition and mixture method on the microstructure and mechanical properties of silicon carbide Ceram. Int. 13 3768
[18] Maitre A, Pat A V, Laval J-P, Valette S and Trottier G 2008 Role of boron on the spark plasma sintering of an α-SiC powder J. Eur. Ceram. Soc. 28 1881–90
[19] Lonello F, Bonnefont G, Leconte Y, Herlin-Boime N and Fantozzi G 2012 Processing of nano-SiC ceramics: densification by SPS and mechanical characterization J. Eur. Ceram. Soc. 32 633–41
[20] Clegg W J 2000 Role of carbon in the sintering of boron-doped silicon carbide J. Am. Ceram. Soc. 83 1039–43
[21] Hammingr R 1989 Carbon inclusions in sintered silicon carbide J. Am. Ceram. Soc. 72 1741–4
[22] Prochazka S and Scanlan R M 1975 Effect of boron and carbon on sintering of SiC J. Am. Ceram. Soc. 58 72–72
[23] Van Rijswijk W and Shanelled D I 1990 Effects of carbon as a sintering aid in silicon carbide J. Am. Ceram. Soc. 73 148–9
[24] Tanaka H 2011 Silicon carbide powder and sintered materials J. Ceram. Soc. Jpn. 119 218–33
[25] Stobierski L and Gubert A 2003 Sintering of silicon carbide. Effect of carbon Ceram. Int. 29 287–92
[26] Magnani G, Sico G, Brentari A and Fabbri P 2014 Solid-state pressureless sintering of silicon carbide below 2000°C J. Eur. Ceram. Soc. 34 4095–8
[27] Bolog M, Sedlackova K, Zifak P and Janega J 2005 Liquid phase sintering of SiC with rare-earth oxides Ceramics- Silikaty 49 259–62
[28] Rçceka M, Gorný G, Stobierski L and Rożniewski K 2001 Effect of carbon content on the microstructure and properties of silicon carbide-based sinters Mater. Charact. 46 245–9
[29] Lanfant B, Leconte Y, Bonnefont G, Garnier V, Jurand Y, Le Gallet S, Pinault M, Herlin-Boime N, Bernard F and Fantozzi G 2015 Effects of carbon and oxygen on the spark plasma sintering additive-free densification and on the mechanical properties of nanostructured SiC ceramics J. Eur. Ceram. Soc. 35 3369–79
[30] Guillard F, Allemain A, Lukivicz J-D and Galy J 2007 Densification of SiC by SPS—effects of time, temperature and pressure J. Eur. Ceram. Soc. 27 2725–8
[31] Quanshi J, Haijun Z, Suping L and Xiaolin J 2007 Effect of particle size on oxidation of silicon carbide powders Ceram. Int. 33 309–13
[32] Roy J, Chandra S, Das S and Maitra S 2014 Oxidation behaviour of silicon carbide: a review Reviews on Advanced Materials Science 38 29–39
[33] Evans J, Giuliani F and Lee B 2014 Pressureless Sintering of Boron Carbide Imperial College London Department of Materials Science and Engineering (London)
[34] Stobierski L and Gubert A 2003 Sintering of silicon carbide: II. Effect of boron Ceram. Int. 29 355–61
[35] Wachtman J B, Cannon W R and Matthewson MJ 2009 Mechanical Properties of ceramics(Wiley)
[36] Ziccardi C and Haber R 2009 Addition of excess carbon to SiC to study its effect on silicon carbide (SiC) armor, advances in ceramic armor II, ceramic engineering and science proceedings Cocoa Beach 314 97