Microstructure and properties of nano-laminated $\text{Y}_3\text{Si}_2\text{C}_2$ ceramics fabricated via *in situ* reaction by spark plasma sintering

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Abstract: A nano-laminated $\text{Y}_3\text{Si}_2\text{C}_2$ ceramic material was successfully synthesized via an *in situ* reaction between YH_2 and SiC using spark plasma sintering technology. A MAX phase-like ternary layered structure of $\text{Y}_3\text{Si}_2\text{C}_2$ was observed at the atomic-scale by high resolution transmission electron microscopy. The lattice parameters calculated from both X-ray diffraction and selected area electron diffraction patterns are in good agreement with the reported theoretical results. The nano-laminated fracture of kink boundaries, delamination, and slipping were observed at the tip of the Vickers indents. The elastic modulus and Vickers hardness of $\text{Y}_3\text{Si}_2\text{C}_2$ ceramics (with 5.5 wt% Y_2O_3) sintered at 1500 °C were 156 and 6.4 GPa, respectively. The corresponding values of thermal and electrical conductivity were $13.7 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$ and $6.3 \times 10^5 \text{ S} \cdot \text{m}^{-1}$, respectively.

Keywords: $\text{Y}_3\text{Si}_2\text{C}_2$; rare earth silicide carbides; spark plasma sintering (SPS); ternary layered structure ceramic; properties

1 Introduction

Rare earth silicide carbides ($\text{RE}_3\text{Si}_2\text{C}_2$, RE = Y, La–Nd, Sm, Gd–Tm) belong to a new group of ternary layered structure materials, which were first developed by Gerdes *et al.* [1,2]. The crystal structure of these compounds shows an orthorhombic subcell and consists of at least two different superstructures [2]. All the $\text{RE}_3\text{Si}_2\text{C}_2$ compounds were reported to have metallic conductivity and their magnetic ordering

temperatures are lower than 60 K [1]. $\text{Y}_3\text{Si}_2\text{C}_2$ is one of a typical representative member of the $\text{RE}_3\text{Si}_2\text{C}_2$ group. In the $\text{Y}_3\text{Si}_2\text{C}_2$ structure, the *c* axis of the subcell is doubled, and thus it crystallizes in the body-centered orthorhombic system with space group *Imma* (No. 74) [3]. On the other hand, in $\text{Y}_3\text{Si}_2\text{C}_2$ structure, Y atoms form two-dimensionally arranged infinite sheets of edge-sharing octahedra containing C_2 pairs, wherein zig-zag chains of Si atoms are interleaved [2].

Even though the crystal structure of $\text{Y}_3\text{Si}_2\text{C}_2$ is different than hexagonal structure (space group *P6₃/mmc*) of typical MAX phases (where M is an early transition metal, A is an A-group element, and X is carbon or nitrogen [4]), both phases have similar

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characteristics of layered structure and anisotropic bonding. Therefore, they both exhibit anisotropic electrical conductivity, anisotropic mechanical properties, and a low shear deformation resistance [3,5–7]. Zhou *et al.* [3] theoretically predicted that the bulk modulus and shear modulus of $Y_3Si_2C_2$ are 93 and 50 GPa, respectively. Moreover, it was concluded that it is a soft ceramic material (Vickers hardness of 6.9 GPa) with good damage tolerance, due to the low shear deformation resistance and low Pugh's ratios ($G/B = 0.537$, where G is the shear modulus and B is the bulk modulus). Furthermore, the calculated volume expansion upon oxidation of $Y_3Si_2C_2$ was found to be ~26%, which could potentially lead to the sealing of the cracks between silicon carbide fibers (SiC_f) and SiC matrix. Therefore, $Y_3Si_2C_2$ may be a promising interphase material for SiC fiber-reinforced SiC matrix (SiC_f/SiC) composite, because of its easy cleavage, low shear deformation resistance, and low volume expansion upon oxidation [3].

On the other hand, $Y_3Si_2C_2$ is inert when in contact with SiC at temperatures up to 1560 °C, while a liquid phase can be formed at temperatures above 1560 °C via a ternary eutectic reaction, according to the calculated Y–Si–C ternary phase diagram [8]. Thus, $Y_3Si_2C_2$ was successfully used as the sintering additive for SiC and/or SiC/Al_4SiC_4 systems [9,10]. The presence of a liquid phase not only effectively promotes the densification of SiC and/or SiC/Al_4SiC_4 , but also improves the fracture toughness of ceramics by optimizing the grain boundary structure. Most importantly, $Y_3Si_2C_2$ can decompose to SiC and Y_2O_3 (might act as the sintering additives for SiC) at ~1600 °C. Therefore, $Y_3Si_2C_2$ was also successfully used as a transition phase to achieve the seamless joining of SiC ceramics [11]. The joining mechanism was identified as follows: First, the laminated $Y_3Si_2C_2$ structure was formed by the *in situ* reaction between Y coatings with a thickness of 500 nm and SiC matrix in the joining layer at 1400 °C. When the joining temperature increased to 1900 °C, $Y_3Si_2C_2$ disappeared owing to its decomposition at high temperatures. More recently, high-entropy $RE_3Si_2C_2$ /rare earth oxides with strong electromagnetic (EM) wave absorption capability and wide efficient absorption bandwidth were successfully synthesized. This can significantly broaden the applications potential of $RE_3Si_2C_2$ materials [12]. Due to its layered structure, good oxidation resistance, metallic conductivity as well as excellent EM wave

absorption capability, dense $Y_3Si_2C_2$ ceramics might potentially be used as a stealth material and an on-beam-line higher-order-mode (HOM) load.

Even though $Y_3Si_2C_2$ has been demonstrated as the promising sintering additive and joining material for SiC-based advanced ceramics, the synthesis method and basic properties (besides electrical and magnetic properties) of $Y_3Si_2C_2$ bulk ceramics have not yet been investigated. The only reported technique to synthesize $Y_3Si_2C_2$ bulk ceramics consisted of arc-melting of the cold-pressed pellets containing the mixture of Y, Si, and C, followed by their annealing in an evacuated silica tube for 30 days at 900 °C [1]. This process was found to be extremely time consuming, because Y ingots were used as raw materials and the reaction temperature was as low as 900 °C. Spark plasma sintering (SPS) is an effective consolidation technology, which enables densification of ceramics at lower sintering temperatures and shorter time when compared to conventional methods. This is attributed to the presence of high-density electric current, which can promote mass diffusion [13,14].

Therefore, the $Y_3Si_2C_2$ nano-laminated bulk ceramics was successfully fabricated by the *in situ* reaction via SPS in this study. Furthermore, the phase composition, microstructure, mechanical properties, as well as electrical and thermal conductivities of $Y_3Si_2C_2$ were investigated. Both the Vickers hardness and the elastic modulus of the as-produced $Y_3Si_2C_2$ ceramics were in good agreement with the reported theoretically calculated results.

2 Experimental

2.1 Preparation of $Y_3Si_2C_2$

YH_2 powder (Sinopharm Chemical Reagent Co., Ltd., Shanghai, China) with a purity of 99.5% and a mean particle size of 75 µm, and β -SiC powder (99.5%, Eno Material Co., Ltd., Qinhuangdao, China) with a mean particle size of 0.5 µm, were used as raw materials. For the formation of $Y_3Si_2C_2$, the YH_2 and SiC powders were mixed in a stoichiometric ratio of 3.05:2. The *in situ* reaction sintering process was performed in an SPS furnace (HPD 25/1, FCT systems, Germany) at the temperature range of 1300–1500 °C for 30 min under a uniaxial pressure of 30 MPa in an Ar atmosphere. The heating and cooling rates were 50 °C·min⁻¹. The as-obtained $Y_3Si_2C_2$ ceramics surfaces were polished using the final 1 µm diamond suspension.

2.2 Material characterization

The phase compositions of the samples were identified by X-ray diffraction (XRD, D8 Advance, Bruker AXS, Germany) with Cu K α radiation ($\lambda = 1.5406 \text{ \AA}$) under an operating voltage of 40 kV and a current of 40 mA at a step scan of $0.02^\circ/2\theta$ and a step time of 0.2 s. The quantitative phase composition and lattice parameters of the $\text{Y}_3\text{Si}_2\text{C}_2$ phase were determined by Rietveld refinement using the TOPAS software.

The surface and fracture micromorphologies of the specimens were studied by the scanning electron microscope (SEM, Quanta 250 FEG, FEI, USA), equipped with an energy dispersive spectroscopy (EDS) detector. The phase distributions were characterized by electron back-scattered diffraction (EBSD) using a thermal field emission electron scanning microscope (Verios G4 uc, Thermo Scientific, USA), equipped with an EBSD apparatus operating at 20 kV accelerating voltage. For the EBSD analysis, the samples were polished with the final 1 μm diamond suspension, followed by etching using an ion beam (BIB, TIC 3X, Leica, Germany) for 3 h [15]. The microstructure and phase compositions were investigated by a transmission electron microscope (TEM, TalosTM F200x, Thermo Fisher Scientific, USA) system equipped with EDS system. Thin foils for TEM observations were prepared by focused ion beam (FIB, Auriga, Carl Zeiss, UK) technique.

2.3 Measurement of properties

Apparent density (ρ) of the samples was determined by the Archimedes' method. Elastic modulus was measured using a nanoindentation system (Hysitron PI85, Bruker, USA) on the polished surfaces of the specimens. Hardness of the materials was measured using a Vickers diamond indenter (HVs-1000 Digital Micro Vickers Hardness Tester, Beijing Times Mountain Peak Technology Co., China) under a load of 0.5, 2, and 5 N, respectively. The dwell time at a maximum load was always 10 s. At least 20 indents were measured for each specimen. Electrical resistivity of the samples was determined using a four-probe resistance tester (Cresbox, Napson Co., Japan). The thermal diffusivity coefficient (α) and specific heat capacity (C_p) were measured by laser flash method using a Netzsch LFA 457 apparatus (LFA, NETZSCH-Gerätebau GmbH, Germany). The thermal conductivity (κ , $\text{W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$)

was calculated according to Eq. (1) [16]:

$$\kappa = \alpha \rho C_p \quad (1)$$

3 Results and discussion

Figure 1 shows the XRD patterns of the samples sintered at different temperatures. $\text{Y}_3\text{Si}_2\text{C}_2$ (JCPDS No. 70-2799) was the predominant phase in all materials, while some Y_2O_3 (JCPDS No. 83-0927) as an impurity phase was also detected. Rietveld refinement technique was applied to determine the fundamental parameters. The amount of the predominant $\text{Y}_3\text{Si}_2\text{C}_2$ phase was 88.4, 94.3, and 94.5 wt% for the samples sintered at 1300, 1400, and 1500 $^\circ\text{C}$, respectively. The corresponding amount of the Y_2O_3 phase was 11.6, 5.7, and 5.5 wt%, respectively. Figure 2 shows the typical Rietveld refinement of XRD pattern of the sample sintered at 1500 $^\circ\text{C}$. No significant change in the lattice parameters of $\text{Y}_3\text{Si}_2\text{C}_2$ phase was observed for the samples sintered at different temperatures (Table 1). This indicated that the lattice parameters of the $\text{Y}_3\text{Si}_2\text{C}_2$ phase were not affected by the sintering temperature. Furthermore, the lattice parameters obtained from the Rietveld refinement are in good agreement with those determined by both the experimental measurements [2] and the calculation results [3], as presented in Table 1. The reliability of the refinement was confirmed by the reliability factors (R_{wp}), which were 9.1%, 9.0%, and 8.6% for the samples sintered at 1300, 1400, and 1500 $^\circ\text{C}$, respectively (Table 1). In addition, the crystal parameters refined from XRD of the powder samples were similar to the bulk samples (Table 1 and Fig. 2) sintered at 1500 $^\circ\text{C}$. It indicated that the influence of possible residual stresses on the lattice parameters in bulk samples on the XRD refinement was minimal.

Fig. 1 XRD patterns of the samples sintered at various temperatures by reactive SPS.

Fig. 2 Rietveld refinement of the XRD patterns of the samples synthesized at 1500 °C: (a) bulk and (b) powder.

Table 1 Experimental lattice parameters of $\text{Y}_3\text{Si}_2\text{C}_2$ derived from Rietveld refinement and selected area electron diffraction (SAED) patterns, and their comparison with the calculated and experimental values reported in literature

$\text{Y}_3\text{Si}_2\text{C}_2$		a (Å)	b (Å)	c (Å)	R_{wp}	Ref.
Experimental XRD	Bulk 1300 °C	8.4358	15.6731	3.8637	9.1%	This study
	Bulk 1400 °C	8.4361	15.6727	3.8628	9.0%	
	Bulk 1500 °C	8.4418	15.6771	3.8630	8.6%	
	Powder 1500 °C	8.4494	15.6751	3.8636	8.7%	
Experimental SAED		8.44	15.72	—	—	This study
Experimental		8.426	15.634	3.846	—	[2]
Theoretical		8.4699	15.6971	3.8746	—	[3]

It is believed that the $\text{Y}_3\text{Si}_2\text{C}_2$ phase was formed by the reaction between SiC and Y, which was generated from the decomposition of YH_2 at a low temperature (~ 650 °C) [17], according to Eqs. (2) and (3):

According to Reaction (4), the formation of Y_2O_3 can be attributed to the presence of oxygen, which was introduced into the samples during powder homogenization process or during sintering at high temperatures (as a trace impurity in Ar atmosphere). Since it is expected that the amount of Y decreased with the increasing sintering temperature from 1300 to 1400 °C, the amount of Y_2O_3 impurity decreased from 11.6 to 5.7 wt%, accordingly. However, the amount of Y_2O_3 in the sample sintered at 1500 °C was only slightly lower than that of the sample sintered at 1400 °C. This suggested that the reaction between YH_2 and SiC was almost completed at 1500 °C.

The theoretical density of the samples was calculated by the rule of mixture, taking the actual amount of the individual phases into account [18]. The

theoretical density of $\text{Y}_3\text{Si}_2\text{C}_2$ ($4.547 \text{ g}\cdot\text{cm}^{-3}$) and Y_2O_3 ($5.02 \text{ g}\cdot\text{cm}^{-3}$) was used. The calculated theoretical density of the bulk samples was 4.596, 4.574, and $4.565 \text{ g}\cdot\text{cm}^{-3}$ for the samples sintered at 1300, 1400, and 1500 °C, respectively. Thus, the relative densities of as-obtained ceramics were 98.0% (1300 °C), 99.0% (1400 °C), and 99.5% (1500 °C), respectively. It is clear that the highly dense $\text{Y}_3\text{Si}_2\text{C}_2$ ceramic materials were successfully obtained by the *in situ* solid state reaction in a significantly shorter time when compared to the previous study, in which Y ingots were used [1]. The use of pulsed current sintering probably improved the mass diffusion and promoted the solid-state reaction, which enable densification to be completed in a relatively short period of time [19]. In addition, YH_2 is more stable than Y, which easily oxidizes to Y_2O_3 during the experimental process. The oxidized layer of Y_2O_3 would inhibit the diffusion and reaction between Y and SiC. Thus, the use of YH_2 as a raw material instead of Y probably accelerated the diffusion process, but also decreased the synthesis temperature due to the decomposition of YH_2 at a relatively low temperature (650 °C) [17]. Furthermore, the presence of YH_2 hydride powder may have facilitated the nucleation of

$\text{Y}_3\text{Si}_2\text{C}_2$ [20].

Figure 3 shows the microstructure of the samples sintered at different temperatures, detected by EBSD. Figures 3(a)–3(c) present the diffraction pattern quality quantified using the “band contrast”, while Figs. 3(d)–3(f) show the phase distribution of $\text{Y}_3\text{Si}_2\text{C}_2$ (in red) and Y_2O_3 (in blue). The elongated, plate-like morphology of $\text{Y}_3\text{Si}_2\text{C}_2$ was clearly identified. The phase fraction of $\text{Y}_3\text{Si}_2\text{C}_2$ measured in the observed area increased with increasing sintering temperature: 80% (1300 °C), 84% (1400 °C), and 91% (1500 °C). At the same time, the grain size distribution is shown in Figs. 3(g)–3(i). The

mean grain size of the materials increased from 3.9 μm (1300 °C) to 8.8 μm (1500 °C). The abnormal grain growth was obviously observed when the sintering temperature was increased to 1400 and 1500 °C.

Figures 4(a)–4(c) show the fracture surfaces of $\text{Y}_3\text{Si}_2\text{C}_2$ sintered at 1300, 1400, and 1500 °C, respectively. The failure mode was mainly intragranular, because of a low shear deformation resistance of $\text{Y}_3\text{Si}_2\text{C}_2$ [3]. Some pores were observed for the sample sintered at 1300 °C (Fig. 4(a)). On the other hand, almost fully dense $\text{Y}_3\text{Si}_2\text{C}_2$ was observed after sintering at 1400 °C (Fig. 4(b)) and 1500 °C (Fig. 4(c)).

Fig. 3 Microstructures of samples fabricated at various temperatures observed by EBSD. Micrograph in band contrast: (a) 1300, (b) 1400, (c) 1500 °C; phase distribution: (d) 1300, (e) 1400, (f) 1500 °C ($\text{Y}_3\text{Si}_2\text{C}_2$ in red and Y_2O_3 in blue); grain size distribution: (g) 1300, (h) 1400, (i) 1500 °C.

Fig. 4 SEM images of the fracture surfaces of $\text{Y}_3\text{Si}_2\text{C}_2$ sintered by reactive SPS at (a) 1300, (b) 1400, and (c) 1500 °C.

TEM analysis was carried out to observe the atomic-scale microstructure of the $\text{Y}_3\text{Si}_2\text{C}_2$ sintered at 1500 °C. Figures 5(a)–5(e) show a high angle annular dark field (HAADF) image and the corresponding elemental distribution of Y, C, O, and Si, respectively. The semi-quantitative EDS analysis confirmed the presence of $\text{Y}_3\text{Si}_2\text{C}_2$ and Y_2O_3 , which correspond to the points 1 and 2 in Fig. 5(a), respectively. The EDS results are presented in Table 2. The presence of a small amount of oxygen was not confirmed. The

Fig. 5 TEM image of $\text{Y}_3\text{Si}_2\text{C}_2$ ceramics sintered at 1500 °C: (a) HAADF image and elemental distribution of (b) Y, (c) C, (d) O, (e) Si; (f) SAED pattern for the yellow area in (a); (g) high resolution transmission electron microscopy (HRTEM) image of $\text{Y}_3\text{Si}_2\text{C}_2$, the insert includes structure models showing the positions of Y, Si, and C. HRTEM image and corresponding SAED pattern were confirmed along the [001] zone axis.

Table 2 EDS results of the spots 1 and 2 in Fig. 5(a)

Spot number	Composition (at%)				Probable phase
	Y	Si	C	O	
1	48.3	23.2	19.7	8.8	$\text{Y}_3\text{Si}_2\text{C}_2$
2	42.2	0	4.5	53.3	Y_2O_3

atomic-scale microstructure along the [001] zone axis was confirmed by HRTEM and corresponding SAED pattern shown in Figs. 5(f) and 5(g). The layered atomic stacking can be clearly seen in the HRTEM image. The lattice fringe spacing of 0.786 nm can be assigned to the (020) planes of $\text{Y}_3\text{Si}_2\text{C}_2$, as shown in Fig. 5(g). The corresponding SAED pattern also confirmed the orthorhombic crystal structure of $\text{Y}_3\text{Si}_2\text{C}_2$ (Fig. 5(f)). The lattice parameters were derived to be $a = 8.44 \text{ \AA}$ and $b = 15.72 \text{ \AA}$, which are in good agreement with those determined from the XRD pattern (Table 1).

The properties of the as-sintered $\text{Y}_3\text{Si}_2\text{C}_2$ are listed in Table 3, along with the properties of some typical ternary carbides. Both the elastic modulus and the Vickers hardness of the $\text{Y}_3\text{Si}_2\text{C}_2$ materials decreased with increasing sintering temperature. This was probably caused by a decreasing amount of Y_2O_3 in the materials with increasing temperature. The elastic modulus and Vickers hardness of Y_2O_3 are ~180 and 7.6 GPa [21–23], respectively, which are slightly higher than the calculated values for $\text{Y}_3\text{Si}_2\text{C}_2$ [3]. Moreover, the grain size increased with increasing sintering temperature, and thus the Vickers hardness also decreased with the increase in the sintering temperature according to the Hall–Petch relationship. The elastic modulus of the sample sintered at 1500 °C was close to the calculated values reported by Zhou *et al.* [3]. The Vickers hardness of the sample sintered at 1500 °C was 7.2 ± 0.8 , 6.5 ± 0.5 , and 6.4 ± 0.4 GPa for the indentation load of 0.5, 2, and 5 N, respectively. These values are in good agreement with the reported calculated value of 6.9 GPa [3].

The shape of Vickers indents was irregular with the exfoliated surfaces and deformed particles, which is like the typical indent shape of Ti_3SiC_2 MAX phase [24]. A typical surface morphology at the tip of a Vickers indent for the sample sintered at 1500 °C is shown in Figs. 6(a) and 6(b). Interestingly, in the case of basal $\text{Y}_3\text{Si}_2\text{C}_2$ plane oriented parallel to the indentation load, typical nano-laminated fracture was observed, owing to the kink boundaries, delamination, and slipping (Fig. 6(a)). Such behavior is commonly

Table 3 Density, mechanical, thermal, and electrical properties of the as-obtained $\text{Y}_3\text{Si}_2\text{C}_2$, and their comparison with the reported values of typical ternary layered structural ceramics

Property	$\text{Y}_3\text{Si}_2\text{C}_2$			Theoretical [1,3]	YAl_3C_3 [25]	Ti_3SiC_2 [4]	Ti_3AlC_2 [26]
	1300 °C	1400 °C	1500 °C				
Theoretical density (g/cm^3)	4.504	4.528	4.543	4.547	3.90	4.52	4.2
Elastic modulus (GPa)	185	177	156	127	325	322	297
Vickers hardness (GPa)*	6.9	6.7	6.4	6.9	12.6	4	3.5
Thermal conductivity ($\text{W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$)	16.1	15.7	13.7	N/A	N/A	37	40
Electrical resistivity ($10^{-6}\ \Omega\cdot\text{cm}$)	132.2	138.2	159.1	100–270	N/A	22.7	38.7

* The value of the Vickers hardness was measured with the load of 5 N.

Fig. 6 Surface morphologies of sample sintered at 1500 °C after Vickers indentation test: (a) basal $\text{Y}_3\text{Si}_2\text{C}_2$ plane oriented parallel to the load with the presence of kinks, delamination, and slipping; (b) basal $\text{Y}_3\text{Si}_2\text{C}_2$ plane oriented perpendicular to the load showing the exfoliation; and (c) high magnification SEM image showing the nano-laminated structure of $\text{Y}_3\text{Si}_2\text{C}_2$.

observed for the group of MAX phases, which belong to typical damage tolerant ceramics [27]. On the other hand, when the basal plane of $\text{Y}_3\text{Si}_2\text{C}_2$ was oriented in a direction perpendicular to the indentation load, the exfoliation and sharp steps-like fracture caused by crack deflection inside the $\text{Y}_3\text{Si}_2\text{C}_2$ grains were observed (Fig. 6(b)). The fracture energy can be consumed by virtue of crack deflection. A typical nano-laminated MAX phase-like structure of $\text{Y}_3\text{Si}_2\text{C}_2$ is shown in Fig. 6(c), which can be easily recognized by its cleavage nature. A relatively low Vickers hardness and typical nano-laminated fracture behavior indicated that $\text{Y}_3\text{Si}_2\text{C}_2$ belongs to the group of soft ceramics. Zhou *et al.* [3] reported that the low shear deformation resistance along the (010) [101] slip system could be attributed to the weak metallic bonding between $\text{Y}_2\text{--C}$.

Thermal conductivity of the $\text{Y}_3\text{Si}_2\text{C}_2$ samples decreased from 16.1 to 13.7 $\text{W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$ with the increase in the sintering temperature from 1300 to 1500 °C. This was observed despite the fact that the grain size increased with increasing sintering temperature (Fig. 3), which usually leads to the improved thermal conductivity due to the decreased phonon scattering by the grain boundaries. Therefore, the decreased thermal conductivity with increasing sintering temperature in this study can be attributed to the presence of Y_2O_3 ($27\ \text{W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$) [28], whose content also decreased with increasing

sintering temperature.

The electrical resistivity of the samples is presented in Table 3. The corresponding electrical conductivity of the samples sintered at 1300, 1400, and 1500 °C were 7.6×10^5 , 7.6×10^5 , and $6.3\times 10^5\ \text{S}\cdot\text{m}^{-1}$, respectively. The electrical conductivity of $\text{Y}_3\text{Si}_2\text{C}_2$ was mainly affected by $\text{Y}1\ 4d_{eg}$, $\text{Y}2\ 4d_{t2g}$, $\text{C}\ 2p_x$, and $\text{C}\ 2p_z$ states (x' and z' are inclined to the x and z axis, respectively, at about 45°), according to the analysis of the projected density of states and the decomposed distribution of electron density [3].

4 Conclusions

The highly dense $\text{Y}_3\text{Si}_2\text{C}_2$ ceramic material was fabricated by *in situ* solid state reaction between YH_2 and SiC via SPS. The as-obtained $\text{Y}_3\text{Si}_2\text{C}_2$ ceramic exhibited a nano-laminated structure, which was confirmed by HRTEM analysis. The lattice parameters were derived as $a = 8.4418\ \text{\AA}$, $b = 15.6671\ \text{\AA}$, and $c = 3.863\ \text{\AA}$ by the Rietveld refinement of XRD patterns. The experimentally measured elastic modulus (156 GPa) and Vickers hardness (6.4 GPa) of the fabricated ceramics are in good agreement with the reported theoretically calculated values. Typical nano-laminated fracture

behavior was observed at the tip of Vickers indents, which indicated that $\text{Y}_3\text{Si}_2\text{C}_2$ belongs to the group of soft ceramics. The thermal and electrical conductivity of the sample sintered at 1500 °C was $13.7 \text{ W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$ and $6.3\times 10^5 \text{ S}\cdot\text{m}^{-1}$, respectively. The proposed synthesized strategy could potentially be used to fabricate other $\text{RE}_3\text{Si}_2\text{C}_2$ phases.

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