Electric conductivity and microstructure evolution of polymer-derived SiAlCO ceramics

Yejie Cao, Xueping Yang, Linan An*

Department of Materials Science and Engineering, Advanced Materials Processing and Analysis Center, University of Central Florida, Orlando, FL 32816, USA

Abstract

The electric conductivity and microstructure of polymer-derived SiAlCO ceramic were studied. The conductivity increased drastically with increasing pyrolysis temperature and exhibited a typical Arrhenius dependence on pyrolysis temperature with the activation energy of $\sim 7.15\text{eV}$. The microstructure was analyzed by XRD, Raman spectroscopy and XPS. Unlike previous reported PDCs, the increase in the conductivity of the SiAlCO cannot be explained by the structural changes of the free carbon phase. It is speculated that the increased conductivity is possibly due to the redistribution of C-O bonds within the free carbon phase.

Key words: polymer-derived ceramic; SiAlCO; electric conductivity; activation energy

* Corresponding author.

I. Introduction

Electric materials that can be used at high-temperatures are of great interests. Polymer-derived ceramics (PDCs), which are directly synthesized by thermal decomposition of polymeric precursors, are one of the promising candidates because of their outstanding multifunctionality, excellent high-temperature properties,¹⁻⁴ as well as flexible manufacturing capability.⁵⁻⁷ Due to their unique structure which consists of amorphous matrix and free-carbon phase, PDCs exhibit many distinct electric properties such as giant piezoresistivity,^{8,9} high-temperature semiconducting behavior.^{10,11} Tremendous attentions are attracted to these electric properties and they have already been explored in various fields such as anodes for lithium ion (Li^+) batteries,¹² electro-thermo-mechanical transducers in MEMS¹³ as well as high-temperature sensors.^{14,15}

Silicon oxycarbide (SiCO) is one kind of PDCs with composition of $\text{SiC}_x\text{O}_{2(1-x)} + y\text{C}_{\text{free}}$, where $\text{SiC}_x\text{O}_{2(1-x)}$ represents the amorphous matrix and C_{free} represents the free-carbon phase.¹⁶ Recently, it is reported that by introducing aluminum into PDC systems, their oxidation and corrosion resistance as well as thermal stability can be significantly improved.¹⁷⁻²⁰ Compared to other systems, the polymer precursors for SiCO ceramic are commercially available at low price, nontoxic and non-sensitive to environment, thus SiAlCO ceramic is receiving more recent attentions. However, most studies are focused on its high-temperature behavior and structure evolution,^{20,21} its electric properties are barely mentioned.

Electrical conductivity is one of the basic electric properties of PDCs, which has been extensively studied. Several conduction mechanisms for PDCs have been proposed. If the free-carbon phase is continuous, the conduction of PDCs is dominated by the glassy carbon phase.²² If the free-carbon phase is isolated, two mechanisms are followed.²³ When the free-carbon

concentration is higher than a certain critical concentration, the conduction is controlled by a tunneling-percolation process; when the free-carbon concentration is lower than the critical concentration, the conduction mechanism is controlled by the Si-based amorphous matrix, following the amorphous semiconducting behavior. Previous studies showed that for polymer-derived SiCN ceramic, the increase in the electric conductivity of the material with increasing pyrolysis temperature is due to the increase in conductivity of free-carbon phase.²⁴ However for the acrylic acid-modified SiCN, the conduction mechanism could not be explained clearly.²⁵ These phenomena suggest that the existence of oxygen highly affect the conductivity of the materials. Since a large amount of oxygen also exists in the SiAlCO ceramic, its conduction behavior could be more complex. Therefore in this paper, the electric conductivity and related microstructure evolution of SiAlCO were investigated.

II. Experimental procedure

The starting materials used in this study are commercially available silicone resin (SIRLES® H62C, Wacker Chemical Corporation, Adrian, MI) and aluminum tri-sec-butoxide (ATB, $(C_4H_9O)_3Al$, Sigma-Aldrich, St. Louis, MO). 5 wt% of aluminum tri-sec-butoxide was first mixed with the silicone resin by using ethanol (C_2H_5OH , Sigma-Aldrich, St. Louis, MO) as solvent. The mixture was continually stirred at room temperature until the ATB was completely dissolved in the silicone. The liquid mixture was then solidified at 150°C for 18 hours in air and followed by a heat-treatment at 350°C for another 4 hours in 100% ultrahigh purity argon (Ar, Airgas, Radnor, PA). The obtained solid was ball milled and sieved to obtain powder of ~1 μm . The powder was compressed at a uniaxial pressure of 20 MPa followed by a cold isostatic pressure of 200 MPa to form green bodies. These green bodies were pyrolyzed into amorphous

SiAlCO ceramics at 1000, 1100, 1200, 1300 and 1400°C for 4 hours in a tube furnace under ultrahigh purity argon, respectively.

To measure the electric conductivity, silver paste was painted on both sides of samples as electrodes. The DC conductivity was obtained by measuring samples' I-V curves on KEITHLEY 2400 (Keithley Instruments, Inc., Cleveland, OH). X-Ray Diffraction (XRD) was recorded on an X-ray Diffractometer (XRD Rigaku, Tokyo, Japan). Raman spectra were obtained using a Renishaw inVia Raman microscopy (Renishaw Inc., Gloucestershire, UK) with the 514nm line of silicone-solid laser as the excitation source. X-Ray Photoelectron Spectroscopy (XPS) was carried out on an AXIS ULTRA (Kratos) apparatus (PHI-5400, Perkin Elmer, Waltham, U.S.A), using monochromatic Al K α radiation ($h\nu=1486.6\text{eV}$) at 150W. To remove possible contaminations, the surfaces of the samples were first etched by Ar⁺ ions for 10 min. The spectra were recorded at room temperature under high vacuum (10^{-9} Torr). The chemical composition of the SiAlCO was determined by a combination of IR carbon-sulfur analyzer and oxygen-nitrogen analyzer (EMIA-320V/EMGA-620V, Horiba Co., Hakata-ku, Japan) to be SiC_{0.93}O_{0.57}Al_{0.03}, which is similar for the samples pyrolyzed at different temperatures.

III. Results and Discussion

Figure 1 shows the XRD spectra of the obtained SiAlCO ceramics. It is seen that all samples are amorphous without detectable diffraction peak, expecting that pyrolyzed at 1400°C which shows a very weak peak corresponding to SiC. The results suggest that the material is highly stable against crystallization.

The measured DC conductivities of the samples pyrolyzed at different temperatures are listed in Table 1. The conductivities increase by ~6 orders of magnitude when the pyrolysis

temperature increases from 1000 to 1400°C, which is much larger than the conductivity change of SiCN and SiC ceramics reported previously. Figure 2 is the plot of the natural logarithm of conductivity $[\ln(\sigma)]$ versus the reciprocal of pyrolysis temperature $[1000/T]$, it shows the conductivity follows the Arrhenius law, according to equation:

$$\sigma = \sigma_0 \exp\left(-\frac{\mu}{kT}\right) \quad (1)$$

where σ is the conductivity of the SiAlCO ceramics, μ is activation energy, T is pyrolysis temperature, k is Boltzmann constant and σ_0 is a prefactor, the calculated activation energy μ equals to 7.15eV. Previously, Chen²⁶ *et al* has reported if the increase in conductivity is due to the free-carbon phase within the material, the activation energy obtained from the Arrhenius relation should be similar to that for the sp^3 -to- sp^2 transition of the free-carbon phase, which is between 3.3 to 3.6 eV.²⁷⁻²⁹ The difference between activation energies suggests the increase in conductivity is not only depended on the free-carbon phase.

In order to have a better understanding of the conduction mechanism, Raman spectroscopy was carried out to analyze the structure of SiAlCO ceramics. Figure 3 is the Raman spectra of SiAlCO samples pyrolyzed at different temperatures. The spectra contain the typical features of free-carbon in polymer-derived ceramics: D peak at $\sim 1350 \text{ cm}^{-1}$ due to the breathing modes of sp^2 carbon atoms in rings and G peak at $\sim 1600 \text{ cm}^{-1}$ resulting from in-plane bond stretching of sp^2 carbon.³⁰⁻³⁴ The Raman spectra were curve-fitted by using Lorentzian function for D peak and Breit–Wigner–Fano function³⁵ for G peak. The position and full width at half maximum (FWHM) were obtained from the curve-fitting, and plot of position and FWHM of G peak as a function of pyrolysis temperature is shown in Figure 4a and 4b, respectively. It is seen,

with increasing pyrolysis temperature, the position of G peak increases from 1585 cm⁻¹ to 1606 cm⁻¹ and the FWHM decreases from 112 to 52.

The activation energy for the free carbon crystallization can be calculated according to the equation proposed by Ferrari *et al.*:²⁹

$$E = kT_c \ln(v\tau) \quad (2)$$

where k is Boltzmann constant, v is the phonon frequency of carbon which is $\sim 5 \times 10^{13}$, τ is a constant about ~ 1 s, and T_c is a characteristic temperature where the structure of the amorphous carbon exhibits a sudden change. However, unlike previous reported polymer-derived SiC²⁴ and SiCN,²⁶ there is no sudden change in either peak position or FWHM.

The size of the free carbon clusters can also be obtained by using the intensity ratio of the D-to-G peaks according to the Ferrar-Robertson equation:^{32,36}

$$\frac{I_D}{I_G} = C'(\lambda) * L_a^2 \quad (3)$$

where $C'(\lambda) = C_0 + \lambda C_1$, ($C_0 = -12.6$ nm and $C_1 = 0.033$), and λ is the wavelength of the excitation source (for $\lambda = 514$ nm). The carbon size as a function of pyrolysis temperature is plotted in Figure 5. The carbon size decreases with increasing pyrolysis temperature at first, then increases with increasing pyrolysis temperature, this phenomenon can be clearly interpreted by Ferrari Model^{32,34} that the decrease in the size at first is due to the sp^3 -to- sp^2 transition and rearrangement of distorted aromatic rings to six-membered rings; and then the increase in the size at higher temperatures is due to the in-plane growth of nano-polycrystalline graphite. In addition, a sudden change point around 1100°C can be observed from Figure 5, assuming

1100°C is the characteristic temperature T_c , the activation energy for the graphitization of free-carbon can be calculated as 3.73 eV, which is consistent with the reported value for crystallization of amorphous carbon,²⁷⁻²⁹ but much smaller than that of the conduction change, suggesting the change in conductivity of the SiAlCO is not due to the carbon crystallization.

X-ray photoelectron spectroscopy (XPS) was used to further analyze the free-carbon phase within the material. Fig 6 is a typical XPS spectrum of the C1s electron state for the SiAlCO ceramic prepared at 1100°C. The complex spectrum can be deconvoluted in four peaks, the peaks at 282.7eV, 284eV, 285eV and 286.5eV are assigned to C-Si,³⁷ C=C,³⁸ C-C/H³⁷ and C-O³⁹ bonds, respectively. The relative concentration of the four bonds can be estimated from the ratio of the integrated area under each peak and the results are also listed in Table 1. It can be seen that the relative concentrations of C-Si and C-O bonds almost remain the same regardless of pyrolysis temperature. Additional, the relative concentration of C=C bond increases with the increasing pyrolysis temperature and the relative concentration of C-C/H bond decreases with the increasing pyrolysis temperature.

The sp^3 -to- sp^2 transition can be described by the following reaction:

and the equilibrium equation of the reaction is

$$\frac{[sp^2]}{[sp^3]} = K = K_0 \exp\left(-\frac{\mu}{kT}\right) \quad (5)$$

where $[sp^3]$ and $[sp^2]$ are the concentrations for sp^3 and sp^2 , respectively; K is the equilibrium constant, K_0 is a coefficient, k is the Boltzmann constant, and μ is the activation energy. Figure 7

is the plot of the natural logarithm of $[sp^2]/[sp^3]$ as a function of pyrolysis temperature $[1000/T]$. It also follows the Arrhenius relation and the activation energy of this reaction equals to 0.5635eV. This value is much smaller than the activation energy for sp^3 -to- sp^2 transition of the free-carbon phase, which is 3.3-3.6eV.²⁷⁻²⁹

All the analytical results above suggest that the conduction mechanism of the increase in the conductivity with increasing pyrolysis temperature of the SiAlCO could be very complex. According to previous studies, if the change of conductivity of PDC attributes to the change of the conductivity of its free-carbon phase, the activation energy for the free-carbon graphitization and sp^3 -to- sp^2 transition should be close to that of the conductivity change.^{24,40} For the SiAlCO, the activation energy for the conductivity change is not consistent with that for the graphitization of the free-carbon phase; meanwhile, the activation energy for sp^3 -to- sp^2 transition of free-carbon is also much smaller than that for the conductivity change. These indicate the change of conductivity of SiAlCO is not only dominated by the structural change in the free-carbon phase.

Besides the effect of the structural evolution of the free-carbon phase, another factor that could affect the conductivity of the SiAlCO is the concentration of oxygen which forms C-O bond within the free carbon and/or for a layer between the matrix phase and the free-carbon phase. It is well known that the conductivity of graphite strongly depends on oxygen concentration when oxygen acts as dopants within the graphite.⁴¹ On the other hand, the C-O layer between the matrix and free carbon phase can strongly isolate the electron transport from the matrix to the free carbon phase, increasing the overall resistivity. XPS spectra reveal that around 10% of C is bonded with O within the material. Although the relative concentration of C-

O bond did not change with pyrolysis temperatures, the rearrangement of these bonds, such as from inside of the free carbon to the interlayer or reverse, with increasing pyrolysis temperature is highly possible. Thereby, we speculate that the change in the conductivity of the SiAlCO is possibly due to the redistribution of the oxygen within the material, which led to the increase in the conductivity of the free carbon phase.

IV. Summary

The electric conduction behavior and microstructure of polymer-derived SiAlCO ceramics were studied. The conductivity of the material measured increased by ~ 6 orders of magnitude when the pyrolysis temperature increased from 1000°C to 1400°C, and the activation energy calculated from the Arrhenius relation between conductivity and pyrolysis temperature was 7.15eV, which was much higher than that of any reported PDC systems. The structure evolution of the free carbon phase was investigated by XRD, Raman Spectroscopy and XPS. The results indicated that the change in the conductivity cannot be explained by the changes in the free carbon structure. We speculated that the conductivity change of the SiAlCO may be because the redistribution of C-O bonds within the free carbon phase.

Acknowledgments

This work is financially supported by National Energy Technology Laboratory, Department of Energy of USA (DE-FE0007004); and National Science Foundation of USA (CMMI-0927441 and CMMI-1301099).

Reference:

1. R. Riedel, A. Kienzle, W. Dressler, L. Ruwisch, J. Bill, and F. Aldinger, A silicoboron carbonitride ceramic stable to 2,000°C, *Nature*. 382 (1996) 796–798.
2. P. A. Ramakrishnan, Y. T. Wang, D. Balzar, L. An, C. Haluschka, R. Riedel, and A. Herman, Silicoboron-carbonitride ceramics: a class of high-temperature, dopable electronic materials, *Appl. Phys. Lett.* 78 (2001) 3076–3078.
3. Y. Wang, L. An, Y. Fan, L. Zhang, S. Burton, and Z. Gan, Oxidation of polymer-derived SiAlCN ceramics, *J. Am. Ceram. Soc.* 88 (2005) 3075–3080.
4. R. Riedel, L. M. Ruwisch, L. An, and R. Raj, Amorphous silicoboron carbonitride ceramic with very high viscosity at ultra high temperatures, *J. Am. Ceram. Soc.* 81 (1998) 3341–3344.
5. L. Liew, W. Zhang, L. An, S. Shah, R. Luo, Y. Liu, T. Cross, M. Dunn, V. Bright, J. Daily, R. Raj, and K. Anseth, Ceramic MEMS—new materials, innovative processing and future applications, *Am. Ceram. Soc. Bull.* 80 (2001) 25–30.
6. L. An, W. Xu, S. Rajagopalan, C. Wang, H. Wang, Y. Fan, L. Zhang, D. Jiang, J. Kapat, L. Chow, B. Guo, J. Liang, and R. Vaidyanathan, Carbon nanotube reinforced polymer-derived ceramic composites, *Adv. Mater.* 16 (2004) 2036–2040.
7. S. Sarkar, A. Chunder, W. Fei, L. An, and L. Zhai, Superhydrophobic mats of polymer-derived ceramic fibers, *J. Am. Ceram Soc.* 91 (2008) 2751–2755.
8. L. Zhang, Y. Wang, Y. Wei, W. Xu, D. Fang, L. Zhai, K.C. Lin, and L. An, A silicon carbonitride ceramic with anomalously high piezoresistivity, *J. Am. Ceram Soc.* 91 (2008) 1346–1349.

9. K. Terauds, P. E. Sanchez-Jimenez, R. Raj, C. Vakifahmetoglu, and P. Colombo, Giant piezoresistivity of polymer-derived ceramics at high temperatures, *J. Eur. Ceram. Soc.* 30 (2010) 2203–2207.
10. Y. Wang, T. Jiang, L. Zhang, and L. An, Electron transport in polymer-derived amorphous silicon oxycarbonitride ceramics, *J. Am. Ceram. Soc.*, 92 (2009) 1603–1606.
11. H. Y. Ryu, Q. Wang, and R. Raj, Ultrahigh-temperature semiconductors made from polymer-derived ceramics, *J. Am. Ceramic. Soc.* 93 (2010) 1668–1676.
12. D. Ahn, R. Raj, Cyclic stability and C-rate performance of amorphous silicon and carbon based anodes for electrochemical storage of lithium, *J. Power Sources.* 196 (2011) 2179-2186.
13. L.A. Liew, R.A. Saravanan, V.M. Bright, M.L. Dunn, J.W. Daily, and R. Raj, Processing and characterization of silicon carbon-nitride ceramics: application of electrical properties towards MEMS thermal actuators, *Sens. Actuators A.* 103 (2003) 171-181.
14. X. Ren, S. Ebadi, H. T. Cheng, Y. Chen, L. An, and X. Gong, Wireless resonant frequency detection of SiCN ceramic resonator for sensor applications, *IEEE AP-S Int. Symp.* 2011, 1856 -1859.
15. Y. Li, Y. Yu, H. San, Y. Wang, and L. An, Wireless passive polymer-derived SiCN ceramic sensor with integrated resonator/antenna, *Appl. Phys. Lett.* 103 (2013) 163505.
16. G.D. Sorarù, S. Modena, E. Guadagnino, P. Colombo, J. Egan, and C. Pantano, Chemical durability of silicon oxycarbide glasses, *J. Am. Ceram. Soc.* 85 (2002) 1529-1536.
17. Y. Wang, Y. Fan, L. Zhang, W. Zhang, and L. An, Polymer-derived SiAlCN ceramics resist oxidation at 1400°C, *Scr. Mater.* 55 (2006) 295-297.

18. Y. Wang, W. Fei, and L. An, Oxidation/Corrosion of polymer-derived SiAlCN ceramics in water vapor, *J. Am. Ceram. Soc.* 89 (2006) 1079-1082.
19. A. Dhamne, W. Xu, B. Fookes, Y. Fan, L. Zhang, S. Burton, J. Hu, J. Ford, and L. An, Polymer-ceramic conversion of liquid polyaluminasilazanes for SiAlCN ceramic, *J. Am. Ceram. Soc.* 88 (2005) 2415–2419.
20. T. Xu, Q. Ma, and Z. Chen, The effect of aluminum additive on structure evolution of silicon oxycarbide derived from polysiloxane, *Mater. Lett.* 65 (2011) 433-435.
21. T. Xu, Q. Ma, Y. Wang, and Z. Chen, High-temperature behavior of Al-doped polymer-derived SiAlOC glasses in air environment, *Ceram. Int.* 40 (2014) 13787-13792.
22. S. Trassl, M. Puchinger, E. Rössler, and G. Ziegler, Electrical properties of amorphous SiC_xNyHz-ceramics derived from polyvinylsilazane, *J. Eur. Ceram. Soc.* 23 (2003) 781-789.
23. Y. Wang, L. Zhang, W. Xu, T. Jiang, Y. Fan, D. Jiang, and L. An, Effect of thermal initiator concentration on the electrical behavior of polymer-derived amorphous silicon carbonitrides, *J. Am. Ceram. Soc.* 91 (2008) 3971-3975.
24. K. Wang, B. Ma, X. Li, Y. Wang, and L. An, Effect of pyrolysis temperature on the structure and conduction of polymer-derived SiC, *J. Am. Ceram. Soc.* 97 (2014) 2135-2138.
25. X. Li, K. Wang, B. Ma, H. Hong, M. Zhang, J. Liu, Y. Wang, and L. An, Effect of acrylic acid additive on electric conductivity of polymer-derived amorphous silicon carbonitride, *Ceram. Int.* 41 (2015) 7971-7976.
26. Y. Chen, F. Yang, and L. An, On electric conduction of amorphous silicon carbonitride derived from a polymeric precursor, *Appl. Phys. Lett.* 102 (2013) 231902.

27. D.S. Grierson, A.V. Sumant, A.R. Konicek, T.A. Friedmann, J.P. Sullivan, and R.W. Carpick, Thermal stability and rehybridization of carbon bonding in tetrahedral amorphous carbon, *J. Appl. Phys.* 107 (2010) 033523.
28. A.C. Ferrari, B. Kleinsorge, N.A. Morrison, A. Hart, V. Stolojan, and J. Robertson, Stress reduction and bond stability during thermal annealing of tetrahedral amorphous carbon, *J. Appl. Phys.* 85 (1999) 7191-7197.
29. A.C. Ferrari, S.E. Rodil, J. Robertson, and W.I. Milne, Is stress necessary to stabilize sp^3 bonding in diamond-like carbon?, *Diamond Relat. Mater.* 11 (2002) 994-999.
30. E.H. Martins Ferreira, M.V.O. Moutinho, F. Stavale, M.M. Lucchese, R.B. Capaz, C.A. Achete, and A. Jorio, Evolution of the Raman spectra from single-, few-, and many-layer graphene with increasing disorder, *Phys. Rev. B.* 82 (2010) 125429.
31. F. Tuinstra, J.L. Koenig, Raman spectrum of graphite, *J. Chem. Phys.* 53 (1970) 1126-1130.
32. A.C. Ferrari, J. Robertson, Interpretation of Raman spectra of disordered and amorphous carbon, *Phys. Rev. B.* 61 (2000) 14095-14107.
33. A. Sadezky, H. Muckenhuber, H. Grothe, R. Niessner, and U. Pöschl, Raman microspectroscopy of soot and related carbonaceous materials: spectral analysis and structural information, *Carbon.* 43 (2005) 1731-1742.
34. A.C. Ferrari, J. Robertson, Resonant Raman spectroscopy of disordered, amorphous, and diamondlike carbon, *Phys. Rev. B.* 64 (2001) 075414.
35. M.V. Klein, in *Light Scattering in Solids III*, Topics. Appl. Phys. 51 (1982).

36. G.A. Zickler, B. Smarsly, N. Gierlinger, H. Peterlik, and O. Paris, A reconsideration of the relationship between the crystallite size L_a of carbons determined by X-ray diffraction and Raman spectroscopy, *Carbon*. 44 (2006) 3239-3246.
37. J. Binner, Y. Zhang, Characterization of silicon carbide and silicon powders by XPS and zeta potential measurement, *J. Mater. Sci. Lett.* 20 (2001) 123-126.
38. R. Cristescu, G. Socol, I.N. Mihailescu, M. Popescu, F. Sava, E. Ion, C.O. Morosanu, and I. Stamatina, New results in pulsed laser deposition of poly-methyl-methacrylate thin films, *Appl. Surf. Sci.* 208–209 (2003) 645-650.
39. A. Majumdar, G. Das, N. Patel, P. Mishra, D. Ghose, and R. Hippler, Microstructural and chemical evolution of $-CH_3$ -incorporated (low-k) SiCO (H) films prepared by dielectric barrier discharge plasma, *J. Electrochem. Soc.* 155 (2008) D22-D26.
40. Y. Chen, X. Yang, Y. Cao, and L. An, Effect of pyrolysis temperature on the electric conductivity of polymer-derived silicoboron carbonitride, *J. Eur. Ceram. Soc.* 34 (2014) 2163-2167.
41. H.J. Shin, K.K. Kim, A. Benayad, S.M. Yoon, H.K. Park, I.S. Jung, et al, Efficient reduction of graphite oxide by sodium borohydride and its effect on electrical conductance, *Adv. Funct. Mater.* 19 (2009) 1987–1992.

List of Figure Captions

Figure 1: XRD Spectra of polymer-derived SiAlCO ceramics pyrolyzed at different temperatures, as labeled.

Figure 2: A Plot of conductivities as a function of pyrolysis temperature for SiAlCO ceramics.

Figure 3: Raman Spectra of SiAlCO samples pyrolyzed at different temperatures, as labeled.

Figure 4: (a) peak position and (b) the FWHM of G peak as a function of pyrolysis temperature.

Figure 5: A plot of the carbon size as a function of pyrolysis temperature.

Figure 6: A typical XPS spectrum of the C1s electron state for the SiAlCO ceramic prepared at 1100°C.

Figure 7: A plot of the sp^2/sp^3 ratio as a function of pyrolysis temperature.

Table 1: Conductivities and relative bond concentrations of the SiAlCO ceramics pyrolyzed at different temperatures.

Pyrolysis Temperature(°C)	1000	1100	1200	1300	1400
Conductivity(s/cm)	1.33E-7	4.99E-6	3.68E-4	1.90E-2	6.93E-1
C-O	10.45%	10.04%	10.85%	10.33%	
C-Si	21.28%	21.79%	21.15%	21.39%	
C=C	62.10%	63.22%	64.45%	65.83%	
C-C/H	6.15%	4.93%	3.53%	2.42%	

Figure

Figure

Figure

Figure

Figure

Figure

Figure

Figure

Figure