Corrosion behavior of polymer-derived SiHfCN(O) ceramics in salt and acid environments

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Abstract

Powder particles of polymer derived SiHfCN(O) ceramics were pulsed electric current sintered at 1300 and 1500 °C to produce amorphous and partially crystalline ceramic pellets for corrosion studies in salt (NaCl or Na₂SO₄) and acid (HF) environments. While, sodium dramatically accelerated phase transformation and catalyzed the crystallization process, the open porosity acted as the main cause for sodium penetration in these materials. The samples, however, were completely disintegrated during fluoride acid tests. The cristobalite and HfO₂ crystalline phases were severely corroded and it was found that the SiC grains were relatively stable in comparison with other phases in the system.

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1. Introduction

Ceramics are usually more sustainable and significantly resistant to degradation in corrosive environments in contrast to metallic materials. While it is the solubility that usually governs the corrosion behavior in ceramics, it is an electro-chemical process that governs in the case of metals. The corrosion behavior and the corrosion resistance of different classes of ceramic materials, for instance, silicates, oxide and non-oxide ceramics exhibit distinct differences [1,2]. Investigating the resistance of advanced technical ceramics under highly corrosive conditions, such as in strong acids, long-term exposure in salts at elevated temperatures is of immense scientific and technological interest. Such studies will assist in understanding and possible life-time estimates of ceramic components in general for a particular environment [3,4].

Polymer-derived ceramic (PDC) materials have been intensively investigated during the last four decades due to their unique structural properties such as high-temperature stability, oxidation, crystallization and creep resistance [5,6]. The development of Si-containing ternary systems (i.e., silicon oxycarbide (SiOC) and silicon carbonitride (SiCN) ceramics) with improved properties via polymer precursor route was of interest in the last two decades [7–10]. In recent years, quaternary and quinary Si-based systems through PDC route have attracted tremendous interest in this field. The incorporation of additional elements such as B [11–15], Al [16,17], Ti [18], Ta [19], Zr [20–22] or Hf [23–26] to the Si-based polymer offered tunable-properties exemplifying the relative importance of these materials. In recent times, the potential of transition metal oxide incorporated PDCs were explored. Among the transition metal oxides, Hf based materials were believed to be technologically significant for nuclear power plants and electronics industries, due to their high neutron absorption capability and high dielectric constant respectively [27,28]. Synthesis, microstructural, phase transformation and mechanical properties of hafnia incorporated silicon carbonitride ceramic were studied by many researchers [24–26,29]. The oxidation behavior of HfSiCN(O) ceramics for Hf/Si ratio of 0.09 at 1400–1600 °C in atmospheric ambience was recently reported by Terauds et al. [30].

PDCs exhibited immense potential as high-temperature materials for possible use in gas turbine engines or heat-engines where they
can be subjected to multiple hot-corrosion attacks concomitant to concentration of the impurities (i.e., sodium, chlorine and sulfur) in the fuel or air [31,32]. However, the studies on corrosion behavior of PDCs are far less comprehensive and the reliability performances under extreme environmental conditions were found to be limited in comparison to ceramics obtained via other routes. Hence, the present study the hot corrosion studies were carried out in sodium chloride (NaCl) and sodium sulfate (Na₂SO₄) environments. These salts act as a source of corrosive species for sodium, chloride and sulfur. Quite often these laboratory experimental studies are an activated process of the actual corrosion of the materials in that particular environment. Similarly, a critical need exists in the fluoro-chemical manufacturing industries to sustain the corrosive halogenated inorganic acids, especially hydrofluoric acid. Hence, hydrofluoric acid (HF) was used as a source of corrosive species for fluoride exposure. The paper therefore focuses on the corrosion behavior of polymer derived SiHfCN(O) in both amorphous and crystalline states in NaCl, Na₂SO₄ salts and HF acid to evaluate the performance of these ceramics under these environments which were not investigated a priori.

2. Experimental method

2.1. Processing

Hafnium modified polysilazane was synthesized by the chemical modification of polysilazane (HTT1800, Clariant, Germany), followed by the addition of 30 vol% hafnium (IV) n-butoxide (Sigma-Aldrich, India) and mixed for ~30 min. The modified polysilazane was cross-linked at 300 °C for 3 h and subsequently thermolyzed in argon atmosphere at 1300 °C for 2 h at a constant heating rate of 5 °C/min [24]. The as-thermolyzed powder was ground and the powder of particle size less than 125 μm was selected for pulsed electric current sintering (PECS) (Sumitomo Coal Mining Co. Ltd., Japan). The sintering was performed in vacuum at 1300 and 1500 °C temperature. The pressure, holding time and heating rate was maintained constant at 30 MPa, 30 min and 50 °C/min respectively. Two sintered ceramic pellets both in the amorphous (referred as A-1300) and crystalline states (referred as C-1500) of 10 mm in diameter and 2.5 mm in thickness were used for the laboratory corrosion studies. For comparison, unmodified Si–C–N ceramic powder prepared by the thermolysis of polysilazane at 1300 °C in Ar atmosphere was sintered at 1500 °C (referred as SiCN-1500) was also studied.

2.2. Hot corrosion test

The hot corrosion studies were carried out in pure NaCl (mp 801 °C) which act as a source of corrosive species for sodium and chlorine and pure Na₂SO₄ (mp 884 °C) for sulfur. Corrosion is a temperature-assisted activated process with relatively complete wetting occurring at higher temperatures (~1000 °C) under acidic conditions. Hence, test protocols were fixed at 1000 °C for 24 h which accommodate a number of important experimental effects [3,32–35]. The sintered ceramic compacts and the corrosive salts (NaCl or Na₂SO₄) interacted in a covered alumina crucible in which the samples were surrounded by corrosive salt. The as-corroded samples were washed after each corrosion experiment with deionized water to dissolve solid materials deposited on the surface. Samples were weighed before and after salt tests, and no significant weight loss was observed. The apparent densities of these hot corroded ceramic samples were determined by Archimedes’ principle [29].

2.3. Acid corrosion test

The acid corrosion test protocol consisted of keeping the samples in an azeotropic 38.26 HF/61.74 H₂O (wt%) solution at 90 °C (20 °C < Tₑₓₚ < Tₑₓₚ of the azeotrope) for a maximum period of 2 weeks [4]. HF is a weak acid, with a low dissociation constant (Kₐ < 0.15408 nm) in the 2θ range between 20° and 90°. The peaks were indexed with reference to powder diffraction files from the International Centre for Diffraction Data (ICDD) database. The phases of each of the corroded samples were characterized by XRD and SEM coupled with EDS for the observation of the phases formed and morphological changes of the corroded surface. The cross-sectional areas of a corroded sample were also analyzed. For transmission electron microscopy (TEM), the fine powdered samples were dispersed in acetone by ultrasonication for 4 min and a drop of the dispersed solution was deposited on a 3 mm carbon-
coated copper grid using a dropper. Samples were observed in both bright field (BF) and dark field (DF) mode by TEM using TECNAI 20G2, FEI instruments, USA, at an accelerating voltage of 200 kV. The diffraction pattern obtained from TEM was also analyzed and the elemental composition was obtained from the energy dispersive spectroscopy (TEM–EDS).

3. Results and discussion

3.1. Characterization of the sintered samples

The diffractograms of as-sintered A-1300, C-1500 and SiCN-1500 samples are shown in Fig. 1. The XRD results reveal that the A-1300 sample was X-ray amorphous in contrast to C-1500, which had predominantly tetragonal and monoclinic HfO2 crystalline phases. However, the peak observed at $2\theta=26.90^\circ$ for A-1300 was attributed to the possible diffusion of graphite during sintering from the graphite foils used. The high intensity peaks at $2\theta=30.58^\circ$, $35.66^\circ$, $50.92^\circ$, $60.52^\circ$, $70.42^\circ$ and $83.09^\circ$ in the case of C-1500 were assigned to tetragonal HfO2 and the relatively low intensity peaks at $2\theta=28.61^\circ$, $31.74^\circ$, $33.97^\circ$ and $41.87^\circ$ were assigned to monoclinic HfO2. Those below a critical crystalline radius of ~8 nm exist only as a tetragonal phase [25]. The SiC peaks overlapped with t-HfO2 peaks at the scan rate used and limited Si3N4 peaks were also observed in the case of C-1500 sample. Nevertheless, with coarsening the tetragonal phase becomes unstable and transforms to a monoclinic phase, which is the room temperature (RT) stable phase of HfO2.

Phase separation and crystallization (SiC) were obvious in the case of SiCN-1500 as shown in Fig. 1. The absence of Si3N4 phase indicates the unstable nature of Si3N4 in vacuum beyond 1300 °C since the stability of Si3N4 is largely dependent on the partial pressure of nitrogen in the sintering atmosphere. Also, upon sintering beyond 1300 °C, a large drop in the vacuum pressure was detected for the SiCN-1500. The thermodynamic calculations performed earlier on SiCN indicated that at low N2 partial pressures the decomposition of Si3N4 to nitrogen gas and liquid or solid silicon might happen as early as 1300 °C [5]. The SEM images of A-1300 in SE (Fig. 2(a)) and BSE modes (Fig. 2b) revealed a less dense microstructure in contrast to C-1500 which exhibited minimal porosity as can be seen in Fig. 2(c) and (d). The bright regions in the back scattered electron image correspond to the hafnium-rich regions inferring localized variation in hafnium content (in Fig. 2(b) and (d)). Similar microstructural features were observed for the SiHfCN(O) ceramics obtained from the same polymeric precursor (HTT1800) [24] and also in our previous study [23]. The TEM images of powder sample in BF (Fig. 2(e) and DF mode (Fig. 2(f)) revealed a crystallite radius of ~8 nm tetragonal phase (HfO2) and embedded in an amorphous SiCN(O) matrix. TEM elemental composition analysis (Fig. 2(g)) (copper was observed from the grid) provided further support to SEM–EDS analysis (Table 1). The diffraction pattern (Fig. 2(h)) of Fig. 2(e), showed crystalline phases distributed in an amorphous matrix. The elemental composition and density of the as-sintered samples is shown in Table 1. The purpose of this analysis was to give only an idea of the elemental composition, albeit the inaccuracies associated with the quantitative determinations of light elements. The apparent density values for A-1300 and C-1500 were found to be ~2.05 and 2.81 g cm$^{-3}$ respectively (Table 1). It is noteworthy that, the extent of open porosity in these ceramics was significantly reduced from ~19% (A-1300) to ~4% (C-1500). From SEM observations, the average pore size was found to be 20–60 µm (for A-1300) and 5–30 µm (for C-1500). This suggested that the densification of the C-1500 was attributed to increases in sintering temperature. In addition, the presence of Si–O bonds in the Si–C–N(O) matrix enhanced the solid state diffusion in the material system. Moreover, the phase separated HfO2 could have also attributed to the enhanced solid state sinterability [24].

3.2. Corrosion behavior in salts and acid environments

3.2.1. Sodium chloride hot corrosion

The diffractograms of the NaCl corroded A-1300, C-1500 and SiCN-1500 are shown in Fig. 3. The A-1300 sample after hot corrosion transformed from X-ray amorphous to a crystalline state. The sample exhibited cristobalite ($\text{SiO}_2$), sodium silicate ($\text{Na}_2\text{SiO}_3$) and mixtures of monoclinic and tetragonal HfO2 peaks. The low
Fig. 2. SEM images of polished sintered surface of A-1300 (a and b); C-1500 (c and d) samples and TEM images of C-1500 (e and f); EDS analysis (g); diffraction pattern (h).

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Table 1
Elemental composition and density of PECS Si–C–N, SiHfCN(O) in as-sintered and after NaCl, Na2SO4, HF test conditions.

<table>
<thead>
<tr>
<th>System type</th>
<th>Si (%)</th>
<th>Hf (%)</th>
<th>C (%)</th>
<th>N (%)</th>
<th>O (%)</th>
<th>Na (%)</th>
<th>Cl (%)</th>
<th>Density (g cm⁻³)</th>
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<tr>
<td>(as-sintered)</td>
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<tr>
<td>SiCN 1500 °C</td>
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<td>~68</td>
<td>~03</td>
<td>~03</td>
<td>~03</td>
<td>~2.07</td>
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<tr>
<td>SHfCN(O)</td>
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<td>~20</td>
<td>~09</td>
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<td>~2.05</td>
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<tr>
<td>SHfCN(O)</td>
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<td>~21</td>
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<td>SiCN 1500 °C</td>
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<td>~48</td>
<td>~01</td>
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<td>~01</td>
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<tr>
<td>SHfCN(O)</td>
<td>~40</td>
<td>~06</td>
<td>~24</td>
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<tr>
<td>1500 °C (After Na2SO4 salt test)</td>
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<td>~1.2</td>
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<td>~24</td>
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<tr>
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<td>~02</td>
<td>NA</td>
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*~1000 °C for 24 h.

**Samples were completely corroded.

^~90 °C for 2 weeks.

intensity peaks of SiO₂ and Na₂SiO₃ were observed at 21.87° and 27.24° respectively. SiC and sodium oxide (Na₂O) intensity peaks overlapped with the m, t-HfO₂ peaks. The C-1500 sample after hot corrosion transformed from an unstable tetragonal phase to RT stable monoclinic HfO₂ phases. A low intensity peak of Na₂SiO₃ must have formed by the reaction of Na₂O with SiO₂; while Na₂O was formed through the dissolution of hydrated Na (reaction mechanisms are shown through Eqs. 2–4):

\[
2\text{NaCl(l)} + 2\text{H}_2\text{O(g)} \rightarrow \text{Cl}_2(\text{g}) + \text{H}_2\text{O(g)} + 2\text{NaOH(l)}
\]  
(2)

\[
2\text{NaOH(l)} \rightarrow 2\text{Na}_2\text{O(s)} + \text{H}_2\text{O(g)}
\]  
(3)

\[
\text{Na}_2\text{O}_3(s) + \text{SiO}_2(s) = \text{Na}_2\text{SiO}_3(s)
\]  
(4)

The oxygen reactant product (SiO₂) and the sodium attacked product (Na₂SiO₃) are thermally stable phases while the latter is water soluble. The XRD results clearly revealed that the A-1300 sample was in a transition state between X-ray amorphous to that consisting of only stable monoclinic HfO₂ phases and on the other-hand, in C-1500 sample a complete transformation to the stable monoclinic HfO₂ phases was observed. Sodium dramatically accelerated these transformations and catalyzed the crystallization reactions at the heat-treatment temperature. The SiCN-1500 sample in addition to SiC peaks has some low intensity peaks of SiO₂ and Na₂SiO₃ (Fig. 3). It reveals that

SiCN-1500 was influenced by the NaCl hot corrosion, but the attack seemed minimal.

**Mechanism of crystallization:** Sodium or Na₂O activities are associated with large rates of crystallization. According to Eq. 5, sodium breaks oxygen bonds and the sodium in Si–O–Na groups play a vital role and catalyzes the crystallization since it increases the concentration of single-bonded oxygen at the interface.

\[\text{[Na–O–Na]} + \text{[Si–O–Si]} = \text{[Si–O–Si]} + \text{[Na⁺–Na⁺]}\]

According to this kinetic model, the rate of diffusion (μ) should be a function of the stoichiometry of the ceramic and is directly proportional to the concentration of sodium. Nevertheless, the concentration of Na⁺ is not constant at the diffusion interface, since some of the sodium diffuses into the ceramic. This diffusion or spread is deactivated by the rejection of the sodium ahead of the crystallization front and the constant Na⁺ can be given as Eq. 6:

\[C_{\text{Na⁺}} \cdot (Dt)^{1/2} = S\]

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where $D$ is the diffusion coefficient of $\text{Na}^+$ into the ceramic and $t$ is time. The rate of crystallization ($\mu$) can be given as Eq. 7:

$$\frac{dx}{dt} = A C_{\text{Na}^+} = \frac{AS}{(Dt)^{1/2}}$$

(7)

which integrates to $x = B t^{1/2} / D^{1/2}$, if $D$ is independent of time, then $x$ takes this form (Eq. 8):

$$x = kt^{1/2}$$

(8)

This predicts that the observed parabolic behavior of diffusion, where $x$ is the thickness of the crystallized layer and $S$, $A$, $B$, and $k$ are constants at constant temperature [35].

The SE and BSE images of hot corroded A-1300 sample surface and cross-section, are shown in Fig. 4(a)–(c) and for C-1500 samples are shown in Fig. 4(d)–(f). This shows that if the sample has less porosity and density, the attack seems rather limited. The different densification characteristics between the A-1300 and C-1500 samples caused different penetration effects (Fig. 4(c) and (f)). The TEM image of NaCl corroded C-1500 sample, its diffraction pattern and EDS, are shown in Fig. 4(g)–(i). From the TEM image, monoclinic $\text{HfO}_2$ crystals with size of $\sim 24 \text{ nm}$ were observed; similar to what was observed previously [23]. The elemental composition (copper was observed from the grid) determined from TEM–EDS provided further support to SEM–EDS analysis (Table 1). The thickness of the salt penetration region was around $25-30 \mu$m and $10-15 \mu$m for the A-1300 and C-1500 samples respectively. The corrosive sodium species clearly penetrated into the ceramic system and distributed uniformly until the matrix could be detected with an appreciable level again. This sample penetration depth was lower for C-1500 in contrast to A-1300 due to a relatively dense microstructure. Similar characteristics of penetration were also observed in the
hot corrosion of the cordierite ceramics by NaCl [33]. In 1990 Cook et al. developed a model in their study of hot corrosion of ceramic by sodium, sulfur and vanadium molten salts [36]. The ceramic resistance rate is then dependent upon the rate of salt deposition as per the following equation:

\[
\frac{d(M_c)}{dt} = \left( \frac{C}{1-C} \right) \cdot \frac{d(M_s)}{dt}
\]

(9)

where \(M_c\) is the mass of ceramic dissolved, \(A\) is the surface area, \(M_s\) is the mass of salt deposited and \(C\) is the concentration of ceramic in layer. Thus, the salt deposition layer thickness in ceramics can serve as a measure of the corrosion performance for general cases. The corrosion resistance is dependent upon the salt penetration thickness (Eq. 9) and thus, C-1500 sample showed better resistance than A-1300 sample in NaCl. The density of NaCl hot corroded A-1300 and C-1500 samples were ~1.83 and 2.52 g cm\(^{-3}\) respectively. The density decreased after hot corrosion in both the samples indicating chemical attack which was more profound for A-1300. The elemental compositions are given in Table 1 and it was seen that Hf and Si considerably decreased and an expected increase in the O content. Na and Cl were observed in trace amounts.

3.2.2 Sodium sulfate hot corrosion

The A-1300 and C-1500 ceramics were completely hot corroded by Na\(_2\)SO\(_4\) at 1000 °C for 24 h under atmospheric ambient. Corrosion products such as silicon sulfate and silicon sulfide were expected not to form under the experimental conditions at the oxygen pressure of around 1 atm [35]. The attack was proportional to the chemistry of the molten salt sodium sulfate. The major mode of degradation of ceramics under hot corrosion was due to crystallization and associated spallation. The sample ceased to exist after the attack for any further analyses to be carried out.

On the contrary, the SiHfCN(O) system was hot corroded nearer to its melting point of the salt (~900 °C for 24 h) and the corresponding SEM micrographs (Fig. 5(a) and (b)) and the cross-section are shown in Fig. 5(c) and (d). In addition to the visible porosity, the interpenetrating network of porosity/voids also served as a penetration route of corrosive species more deeply into the interior of the corroded sample (see cross-section, Fig. 5(c) and (d)). Both the HfO\(_2\) phases and SiCN matrix were severely attacked in the sodium enriched region. The schematic presentation of this situation for NaCl (~1000 °C for 24 h) and Na\(_2\)SO\(_4\) (~900 °C for 24 h) is exemplified in Fig. 6.

3.2.3 Hydrofluoric acid corrosion

Most of the polycrystalline oxide ceramics exhibit poor resistance in hydrofluoric acid medium [4]. When A-1300 and C-1500 were exposed for a sufficient length of time in aqueous hydrofluoric acid, the samples completely disintegrated resulting in an accumulation of predominantly SiC grains and trace amount of Si\(_3\)N\(_4\) at the bottom of the test vessel. The XRD of the hydrofluoric acid corroded A-1300, C-1500 and SiCN-1500 are shown in Fig. 7. The A-1300 sample after
hydrofluoric acid corrosion showed SiC (hexagonal rhombohedral lattice) peaks at $2\theta = 35.95^\circ$, 60.25$^\circ$ and 72.10$^\circ$. The C-1500 sample after acid corrosion also showed predominantly SiC. In addition to SiC, peak at $2\theta = 26.58^\circ$ were attributed to the graphite during sintering from the foils used. The XRD results revealed that both the samples had predominantly SiC phases and trace amount of Si$_3$N$_4$ after hydrofluoric acid corrosion. It is possible that a small fraction crystalline SiC, Si$_3$N$_4$ phases already existed which might have been masked by the high intensity peaks of HfO$_2$ which were revealed after HF corrosion. Interestingly, non-oxide ceramics SiCN-1500 sample resist the HF acid corrosion (Fig. 7), (all pure carbides have excellent HF acid resistance).

The SEM images of acid corrosion in the initial stages (after 2 days) are shown in Fig. 8(a) and (b). Further, images of corroded A-1300 sample particles and its C-1500 sample particles are shown in Fig. 8(c)–(f) respectively. The SEM observation showed that both the ceramics completely fragmented, but took different time periods for degradation. The former one completely disintegrated after 4–6 days (Fig. 8(c) and (d)) and the latter survived for 10–12 days (Fig. 8(e) and (f)). The individual SiC grains remained corrosion resistant and settled at the bottom of the test vessel. The TEM image of HF corroded C-1500 sample and its EDS, are shown in Fig. 8(g) and (h) (inset image shows diffraction of the Fig. 8(g)). The elemental compositions (Table 1) confirmed the presence of Si and C with trace amounts of O and N.

However, the ceramics containing Si, SiO$_2$, or Si-O based phases are active to corrosion by aqueous hydrofluoric acid. SiO$_2$ is known to dissolve at low temperature (25$^\circ$C) in aqueous hydrofluoric acid, forming hexafluorosilicic acid (e.g., hexafluorosilicic acid) (Eq. 10). The equilibrium reaction product species are SiF$_6$H$_2$(l), which is the preferred species at low temperature. Another possible reaction product is tetrafluorosilicate (Eq. 11) (SiF$_4$(g)) is the equilibrium species.

\[
\text{SiO}_2(s) + 6\text{HF}_1(l) = \text{SiF}_6\text{H}_2(l) + 2\text{H}_2\text{O}_1(l) \quad (10)
\]
\[
\text{SiO}_2(s) + 4\text{HF}_1(l) = \text{SiF}_4(g) + 2\text{H}_2\text{O}_1(l) \quad (11)
\]

It is noteworthy that there is an uncertainty in the available thermodynamic data for the SiO$_2$ reactions [4,37]. The dissolution reactions at 25$^\circ$C for hafnia and silicon carbide are shown in Eqs. 12 and 13.

\[
\text{HfO}_2(s) + 6\text{HF}_1(l) = \text{HfF}_6\text{H}_2(l) + 2\text{H}_2\text{O}_1(l) \quad (12)
\]
\[
\text{SiC}_1(s) + 6\text{HF}_1(l) = \text{SiF}_6\text{H}_2(l) + \text{CH}_4(g) \quad (13)
\]

Hafnia (in SiHfCNO system) reacts with fluoric acid to form fluorohafnate (HfF$_6$H$_2$(l)) [38]. All of the pure carbides possess excellent resistance in the HF acid test and, the Si-based carbides (i.e., SiC) showed excellent corrosion resistance under the hydrofluoric acid test condition in contrast to Si-based nitride ceramics and other Si-containing ceramics. Although SiC can be thermodynamically soluble (see Eq. 13), however, they are more stable in comparison with other phases in this system. Hence, the individual SiC grains remain in the test vessel predominantly.

### 4. Conclusions

Polymer derived SiHfCN(O) ceramics were pulsed electric current sintered at 1300 and 1500 $^\circ$C to make dense ceramic pellets. The corrosion behavior of this SiHfCN(O) ceramics in...
Fig. 8. Hydrofluoric acid corroded (90 °C for maximum of 2 weeks): SEM images of A-1300 (a–d); C-1500 samples (e and f) and TEM image of C-1500 (g); EDS analysis (h) (inset image – diffraction pattern).
sodium, chlorine, sulfur and fluororo environments were studied. The following conclusions are drawn.

4.1. Hot-corrosion by salts

Phase changes in the both the samples were predominantly due to sodium and catalyze the crystallization, in addition to heat-treatment at high temperature. The Na₂SiO₃ phase was formed due to NaCl attack at 1000 °C for 24 h. The different densification characteristics between the samples caused different penetration effects. This shows that the C-1500 sample showed better resistance than A-1300 sample in NaCl. In addition, the ceramics were completely hot corroded by Na₂SO₄ at 1000 °C for 24 h under atmospheric ambient.

4.2. Acid corrosion

This oxide ceramics completely fragmented, resulting in an accumulation of individual SiC grains at the bottom of the vessel. Both the samples were predominantly SiC phases after hydrofluoric acid corrosion. Densification effect causes the disintegration delay in the A-1300, C-1500 system and the one with higher porosity or exposed surface area will exhibit greater corrosion. The SiC grains were more stable in comparison with other phases (SiO₂ and HFO₂) in this system. As a result, the individual SiC grains remain in the test vessel predominantly.

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References


