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# Suppression of $\alpha$ -quartz in montmorillonite–SiOC ceramic nanocomposites with water vapor-assisted pyrolysis†

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Bulk montmorillonite–silicon oxycarbide (MMT–SiOC) ceramic nanocomposites were fabricated with either Ar or Ar–H<sub>2</sub>O pyrolytic atmospheres to determine the effect of water vapor on SiO<sub>2</sub> phase evolution at the MMT–SiOC interface and selectively etched porosity. While water vapor-assisted pyrolysis selectively removes free C in pure SiOC, surprisingly, MMT and water vapor in combination led to encapsulation, shielding, and retention of C as graphene oxide in MMT–SiOC–H<sub>2</sub>O ceramics. The crystalline  $\alpha$ -quartz phase in MMT–SiOC was converted to an amorphous SiO<sub>2</sub> phase during water vapor-assisted pyrolysis, leading to 50–75% reductions in specific surface areas of HF-etched MMT–SiOC without changes to the intrinsic pore structure of water vapor-pyrolyzed SiOC. Further pyrolysis to 1400 °C led to the dissolution of the aforementioned pore structure due to more extensive carbothermal reduction and refinement of  $\beta$ -SiC nanowhiskers. This work provides a new mechanistic understanding of how inorganic 2D fillers (like MMT) can mediate chemical and structural evolution during polymer-derived ceramic synthesis.

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## 1. Introduction

Silicon oxycarbides (SiOCs) are a promising class of polymer-derived ceramics (PDCs) with notable thermal stability, oxidation resistance, and thermomechanical properties that offer attractive applications for aerospace, thermal insulation, catalyst supports, and lightweight structures. Classically amorphous, SiOC glasses are comprised of disordered SiO<sub>x</sub>C<sub>4–x</sub> (1 ≤ x ≤ 3) tetrahedra interspersed with SiO<sub>2</sub>, SiC, and free carbon (C) regimes, in which the composition of Si–O–C tetrahedra may be tailored from the preceramic polymer(s) and pyrolytic temperatures and atmospheres.<sup>1</sup> Inert atmosphere pyrolysis between 800 °C and 1100 °C yields an amorphous SiOC matrix with SiO<sub>2</sub> amorphous domains and turbostratic C regimes spreading between SiOC clusters. Pyrolysis to 1300 °C leads to  $\beta$ -SiC nucleation and growth from carbothermal reduction of amorphous SiO<sub>2</sub> and SiOC, although nuclei sizes are generally restricted to less than 5 nm.<sup>2</sup>

While containing an amorphous SiO<sub>2</sub> phase, SiOC glasses intrinsically exhibit low porosity after selective etching with HF as the SiO<sub>2</sub> phase is either sufficiently small (<2 nm) or shielded by C and SiOC phases serving as diffusional and chemical

barriers.<sup>3</sup> Porous SiOC is generally obtained through sacrificial additives to impart rationally controlled pore structures either through self-foaming during pyrolysis or selective etching of SiO<sub>2</sub>. For PDC systems, sacrificial additives generally must be chemically compatible with preceramic polymers and the evolved ceramics, and maintain coherency with the amorphous/nanostructured SiOC matrix in order to serve as reinforcements rather than inclusions. Promising sacrificial additives for PDCs are then reduced to polymeric fillers or novel oxide nanoparticles, which may involve higher material or production costs.<sup>4</sup> Montmorillonite (MMT) clay additives are popular reinforcements in polymer nanocomposites as a low cost and sustainable filler from natural clay sources, leveraging the intrinsic two-dimensional (2D) nanostructured morphology of ~1 nm layered aluminosilicate sheets to improve mechanical properties and gas permeability, among others.<sup>5</sup> MMT additives are also attractive high-temperature reinforcements due to the inherent stability of the aluminosilicate sheet structure up to 1300–1400 °C before phase separation to cristobalite and mullite.<sup>6</sup>

Novel MMT–SiOC ceramic nanocomposites were recently fabricated with twice-functionalized MMT nanoflakes to achieve an exfoliated 2D nanocomposite microstructure.<sup>7</sup> MMT nanoflakes were stabilized in the SiOC matrix through an interfacial  $\alpha$ -quartz zone originating from the siliceous functional groups on the MMT surface. Selective etching of MMT-derived SiO<sub>2</sub> regimes indicated the emergence of small slit-like pores between 2 nm and 5 nm, but the relatively small specific surface areas (SSAs) of ~100 m<sup>2</sup> g<sup>–1</sup> for MMT–SiOC

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ceramics pyrolyzed to 1200 °C in Ar elucidated that SiO<sub>2</sub> phase evolution from MMT was diffusion limited by excess free/turbostratic C present in MMT-SiOC ceramics. MMT nano-flakes are promising materials for introducing anisotropic pores owing to its 2D morphology but there is a need to improve the intrinsic porosity imbued by MMT structures. Residual  $\alpha$ -quartz also limits the viability of MMT-SiOC in elevated temperature applications due to the  $\alpha$ -to- $\beta$  quartz transition at ~573 °C, during which the notable volumetric transformation can lead to fracture.

Reactive pyrolysis atmospheres (e.g., H<sub>2</sub>O, H<sub>2</sub>, CO) have been utilized to impart additional O, H, or C species and drive phase evolution in SiOC through radical-mediated reactions with Si during the polymer-to-ceramic conversion between 400 °C and 800 °C.<sup>8–14</sup> Notably, water vapor-assisted pyrolysis has been shown to preferentially remove C as CO, suppress SiC formation, and accelerate SiO<sub>2</sub> crystallization in SiOC through an increase of Si–O–Si bonds and a decrease of Si–C bonds.<sup>3,8,15,16</sup> As excess C was evolved as CO gas in the presence of water vapor during the polymer-to-ceramic transition, any turbostratic C retained in the SiOC matrix was comprised of fewer graphitic layers, which diminished the barrier for atomic diffusion. C and O loss was detected by X-ray photoelectron spectroscopy (XPS) as the O<sub>1s</sub>/Si<sub>2p</sub> and C<sub>1s</sub>/Si<sub>2p</sub> peak ratios decreased from 3.997 and 0.625 in SiOC pyrolyzed at 1300 °C in Ar to 3.433 and 0.294, respectively, in SiOC pyrolyzed with Ar + 16.7% H<sub>2</sub>O, with H<sub>2</sub>O injected between 500 °C and 700 °C.<sup>3</sup> These ceramics had compositions of SiO<sub>1.24</sub>C<sub>4.16</sub> and SiO<sub>1.77</sub>C<sub>1.74</sub>, respectively, as determined by elemental analysis that further confirmed that water vapor-assisted pyrolysis volatilized C and minorly O.<sup>3</sup> H<sub>2</sub>O-derived SiO<sub>2</sub> was susceptible to HF etching that produced highly porous SiOC with SSAs of ~2000 m<sup>2</sup> g<sup>−1</sup> and pore sizes between 2 and 5 nm for N<sub>2</sub> sorption experiments.<sup>15</sup> Pore distributions represented the 2–5 nm SiO<sub>2</sub> domains selectively-etched by HF, and N<sub>2</sub> sorption results agreed with transmission electron microscopy (TEM) imaging that revealed the size of both SiO<sub>2</sub> domains and mesopores.<sup>3</sup> Aided by C removal and OH introduction, soft ferromagnetic Ni–SiOC composites produced with water vapor-assisted pyrolysis to 900 °C exhibited magnetic saturations up to 7.14 emu g<sup>−1</sup> and DC conductivities up to 247 S m<sup>−1</sup>.<sup>17</sup> *In situ* metallic Ni and minorly NiO nanoparticles influenced these improvements from Ar + H<sub>2</sub>O pyrolyzed SiOC (<1 emu g<sup>−1</sup>, <50 S m<sup>−1</sup>),<sup>17</sup> which was noteworthy because NiSi<sub>3</sub> predominantly formed in carbon-rich SiOC pyrolyzed in Ar.<sup>18</sup> Water vapor-assisted pyrolysis is an attractive ceramic processing route to catalyze unique phase evolution mechanisms for novel multifunctional SiOC ceramics and composites.

H<sub>2</sub> pyrolysis atmospheres aggressively removed C to not only create visually “white” SiOC powders (compared to optically “black” SiOC from the high free/turbostratic C content) but also imbue photoluminescence from 410 nm to 530 nm with increasing pyrolysis temperatures from 800 °C to 1100 °C.<sup>8,14,15</sup> <sup>29</sup>Si nuclear magnetic spectroscopy (NMR) did not reveal substantial SiOC matrix composition changes after H<sub>2</sub> ingress as only the low concentration of SiOC<sub>3</sub> tetrahedra (<2–3%) inherently present in SiOC was removed by H<sub>2</sub>, but <sup>1</sup>H NMR

uncovered a higher concentration of C–H bonds and the emergence of Si–H bonds.<sup>14</sup> H<sub>2</sub>-assisted pyrolysis was determined to primarily react with the free C phase through decarburization and H<sup>+</sup> implantation. In the presence of CO<sub>2</sub>, Si–C and Si–N bonds were converted to Si–O bonds in SiOC and SiCN systems and reduced the Young's modulus and hardness by ~33%. CO<sub>2</sub> improved ceramic yields and decreased linear shrinkage during pyrolysis through the incorporation of O and was qualitatively evidenced with EDS depth profiles.<sup>13,19</sup> Free C evolved as CO *via* the Boudouard reaction, but these effects were limited to depths of 25–30  $\mu$ m from the surface by limiting CO<sub>2</sub> exposure to less than 800 °C.<sup>13</sup> CO<sub>2</sub>-assisted pyrolysis is generally less investigated than either H<sub>2</sub>- or H<sub>2</sub>O-assisted pyrolysis and has not been appreciably reported for this material system apart from the aforementioned studies.

While reactive H<sub>2</sub> or CO<sub>2</sub> atmospheres were worthy of study, water vapor-assisted pyrolysis was chosen to deconvolute phase evolution mechanisms in MMT-SiOC that may be clouded during typical Ar pyrolysis or intrinsic to SiOC. By isolating the role of MMT in the context of the breadth of water vapor-assisted PDC research, this approach can be leveraged to other 2D nanomaterials such as MXenes (e.g., Ti<sub>3</sub>C<sub>2</sub>T<sub>x</sub>), transition metal dichalcogenides (TMDs), layered double hydroxides (LDHs), or hexagonal boron nitride (h-BN) that have potential as supercapacitors, batteries, electrocatalysts, catalyst supports, and piezoelectric energy harvesters, among others. Inherently an oxide, MMT can be considered as a source for introducing templated porosity through SiO<sub>2</sub> domains derived with Ar + H<sub>2</sub>O pyrolysis. This study focused on the effect of water vapor-assisted pyrolysis on SiO<sub>2</sub> phase formation in MMT-SiOC ceramics to evaluate the role of MMT nanosheet reinforcements as SiO<sub>2</sub> growth sites within the amorphous SiOC matrix (900–1200 °C) and SiC nanowhisker growth at higher pyrolytic temperatures (>1400 °C). Water vapor-assisted pyrolysis further served to selectively remove excess C during the polymer-to-ceramic transition, and the role of C as a diffusional barrier to SiO<sub>2</sub> phase growth was investigated. The MMT content and pyrolytic processing temperature were varied to evaluate selectively etched porosity by N<sub>2</sub> sorption and microstructural features by X-ray diffraction, electron microscopy, and <sup>29</sup>Si NMR to understand phase evolution mechanisms and interfacial characteristics in MMT-SiOC ceramic nanocomposites.

## 2. Experimental methods

### 2.1. Materials

The materials for MMT exfoliation and functionalization were provided in a prior study.<sup>7</sup> A commercial polysiloxane (PSO, Polyamic SPR-684, [−Si(Ph)<sub>2</sub>O−]<sub>3</sub>[−Si(Me)(H)O−]<sub>2</sub>[−Si(Me)(CH=CH<sub>2</sub>)O−]<sub>2</sub>, Starfire Systems, Inc., Glenville, NY) was the base preceramic polymer and cured with a 2.1% platinum-divinyl-tetramethyldisiloxane complex in xylene (Karstedt catalyst, Gelest, Inc., Morrisville, PA). Concentrated hydrofluoric acid (HF, 49 wt%, Avantor Specialty Materials, Inc., Radnor, PA) was used to selectively etch SiO<sub>2</sub> from the MMT-SiOC ceramics. Acetone (technical grade, Fisher Scientific, Waltham, MA), toluene ( $\geq 99.5\%$ , Fisher Scientific, Waltham, MA), and



ultrapure water (NP H<sub>2</sub>O, 18.2 MΩ cm) were used throughout the experiment as general solvents. Argon (Ar, Industrial Grade, Airgas, Christiansburg, VA) was used for the pyrolysis atmosphere. All chemicals were used without further purification.

## 2.2. MMT-SiOC sample preparation and pyrolysis

AEAPTMS-functionalized MMT nanoflakes (AEAPMTS: (3-aminopropyl)trimethoxysilane) were synthesized from a naturally occurring clay source following a prior procedure.<sup>7</sup> xMMT-PSO preceramic green bodies ( $x = 0, 1, 2, 5$  wt%) were prepared by mixing PSO with toluene (~50 v/v% relative to PSO), Karstedt catalyst (100 ppm relative to PSO), and functionalized MMT nanoflakes (100/0, 99/1, 98/2, 95/5 w/w). The suspension was stirred under vacuum to remove the residual solvent. The polymer suspensions were cast into  $\Phi$  12 mm  $\times$  6 mm cylindrical silicone molds, degassed under vacuum, and cured at 60 °C for 6 h, 120 °C for 6 h, and 150 °C for 12 h. The green bodies were lightly ground and edges were slightly beveled to 1500 grit with SiC abrasive pads to remove any mold artifacts. To minimize potential surface oxidation, the green bodies were placed between two sheets of carbon felt in an alumina crucible.

The green bodies were pyrolyzed at 900 °C, 1200 °C, or 1400 °C in a horizontal tube furnace (1830-12 Horizontal Tube Furnace, CM Furnaces, Inc., Bloomfield, NJ), with heating and cooling rates of 2 °C min<sup>-1</sup>, soaking time of 2 h, and inflowing Ar of 500 cm<sup>3</sup> min<sup>-1</sup>. Water vapor was injected between 500 °C and 700 °C by flowing Ar through a water bath maintained at 60  $\pm$  2 °C (corresponding to 80 vol% Ar-20 vol% H<sub>2</sub>O). The temperature was held for 30 min at 500 °C and 700 °C to ensure a thorough changeover of Ar to Ar-H<sub>2</sub>O and Ar-H<sub>2</sub>O to Ar, respectively. To mitigate batch-to-batch variations especially under reactive pyrolytic conditions, a typical furnace run consisted of at least 2 replicates of each xMMT-PSO ( $x = 0, 1, 2, 5$ ) so a complete data set could be compared after each run; at least 3 furnace runs were conducted and samples with starkly different trends in ceramic yields, sample shrinkage, and/or phase profiles from those reported previously and within this study were discarded.

MMT-SiOC ceramics prepared with water vapor-assisted pyrolysis were denoted as xMMT- $T_{\text{Pyr}}$ -H<sub>2</sub>O with  $x = 1, 2, 5$  wt% and  $T_{\text{Pyr}} = 900, 1200, \text{ or } 1400$  °C. Comparable MMT-SiOC ceramics were pyrolyzed without water vapor and denoted as xMMT- $T_{\text{Pyr}}$ -Ar. Blank SiOC ceramics without MMT were similarly denoted SiOC- $T_{\text{Pyr}}$ -Ar or SiOC- $T_{\text{Pyr}}$ -H<sub>2</sub>O.

## 2.3. Characterization

Ceramic yield and volume shrinkage were calculated from changes in mass and volume, respectively, between green bodies and pyrolyzed samples. Reported values were averages of 5–7 intact/unbroken replicates and corresponding errors indicated standard deviations. The phase compositions of powdered samples were analyzed with an X-ray diffractometer (XRD, Empyrean, Malvern PANalytical Ltd, Almelo, The Netherlands) operated at 40 kV and 40 mA with Cu K $\alpha$  radiation between  $2\theta = 3$ –90° and a scan rate of 0.02° s<sup>-1</sup>. Microstructures

were observed with a scanning electron microscope (SEM, JEOL IT-500HR, JEOL USA, Peabody, MA) and a transmission electron microscope (TEM, JEOL 2100, JEOL USA, Peabody, MA) with a LaB<sub>6</sub> filament operated at 200 kV. Single-pulse magic angle spinning (MAS) nuclear magnetic resonance (NMR, Bruker Avance II, Bruker Corp., Billerica, MA) spectra for <sup>29</sup>Si were collected with a frequency of 59.627 MHz, 2.5  $\mu$ s pulse width, 60 s recycle delay, averaged over 5000 scans, and in a 4 mm rotor at a MAS frequency of 10 kHz. <sup>29</sup>Si was calibrated by an external TMS standard (0 ppm). Raman characterization (XploRA PLUS, Horiba Scientific, Piscataway, NJ) was performed with an Ar laser at 532 nm between 500 and 4000 cm<sup>-1</sup>. Elemental analysis (E.A.) for Si (ICP-OES) and C (combustion) was performed by Galbraith Laboratories, Inc. (Knoxville, TN). Young's modulus and hardness were determined by nanoindentation (Nano Indenter G200, Keysight Technologies Inc., Santa Rosa, CA, USA) with a Berkovich diamond indenter to a peak displacement of 1  $\mu$ m and averaged across a 5  $\times$  5 array with a 20  $\mu$ m indent spacing. The samples were polished up to 0.1  $\mu$ m with a diamond suspension prior to analysis. Porosity and surface area characteristics were determined by nitrogen adsorption at 77 K (Quantachrome Autosorb-iQ C-XR, Quantachrome Instruments, Boynton Beach, FL). Samples were degassed under vacuum at 300 °C for 12 h before analysis. xMMT-SiOC ceramics were lightly fractured into ~5 mm pieces and etched in a 10% HF (aq.) solution at 1 w/v% for at least 96 h. Samples were thoroughly washed with NP H<sub>2</sub>O and EtOH and vacuum dried at 120 °C for 12 h. A Non-Local Density Functional Theory (NLDFT) model in the analysis software was applied to the adsorption branch with a spherical/cylindrical pore structure based on a silica adsorbate model to determine pore characteristics.

# 3. Results and discussion

## 3.1. Microstructural and phase evolution

As revealed in a preceding study,<sup>7</sup> exfoliated MMT-SiOC ceramic synthesis was heavily dependent on the MMT nanosheet synthesis and functionalization to not only imbue hydrophobicity and obtain single-to-few-layer nanoflakes, but also mitigate agglomeration or restacking effects between synthesis steps. Single-to-few-layer MMT nanoflakes were first obtained after freeze/thaw/sonication-assisted exfoliation and subsequently functionalized twice with cetyltrimethylammonium bromide (CTAB) and AEAPTMS to impart a covalently-bound organophilic silane coating. The previous study focused on phase and microstructural evolution in such novel MMT-SiOC nanocomposites, in particular noting that nanostructured MMT participated in SiOC-forming reactions during the polymer-to-ceramic transformation that was evidenced by interfacial  $\alpha$ -quartz. This reactivity was leveraged in this study with water vapor-assisted pyrolysis to evaluate MMT as a source of templated porosity.

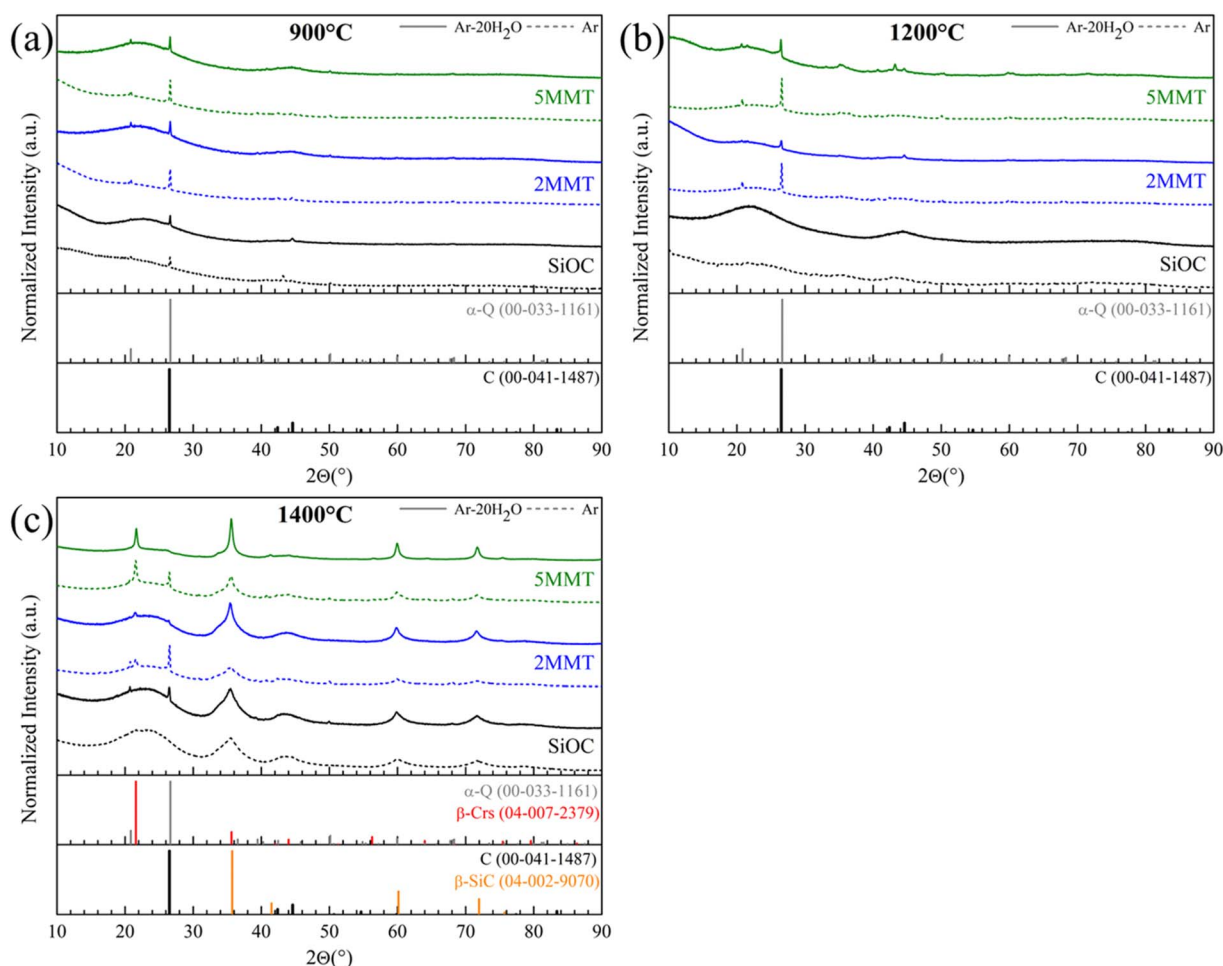
Phase compositions of MMT-SiOC ceramics pyrolyzed in Ar and Ar-20% H<sub>2</sub>O were analyzed with XRD and shown in Fig. 1. 1MMT-SiOC conditions were omitted due to similarities to 2MMT-SiOC conditions, and indexed reference patterns for



identified crystalline phases are provided in the ESI.† SiOCs pyrolyzed with Ar and Ar-20% H<sub>2</sub>O were largely amorphous but with an increased amorphous SiO<sub>2</sub> (am-SiO<sub>2</sub>) halo between 22 and 24° in both SiOC-900-H<sub>2</sub>O (Fig. 1(a)) and SiOC-1200-H<sub>2</sub>O (Fig. 1(b)) samples. While SiOC intrinsically yielded am-SiO<sub>2</sub> at 1400 °C, SiO<sub>2</sub> formation was more pronounced in SiOC-1400-H<sub>2</sub>O. All these observations are consistent with earlier studies,<sup>3,8,15,16</sup> indicating that water vapor pyrolysis accelerated SiO<sub>2</sub> formation and even SiO<sub>2</sub> crystallization in pure SiOC. However, in Fig. 1(a and b), SiOC-900 and SiOC-900-H<sub>2</sub>O samples, along with xMMT-900-Ar and xMMT-1200-Ar ceramics, exhibited crystallized  $\alpha$ -quartz [trigonal; *P*<sub>3</sub>2<sub>1</sub>; PDF 00-033-1161] (100) and (101) reflections at 20.7° and 26.5°, respectively. We believe such  $\alpha$ -quartz formation is transient in pure SiOC, as this phase disappeared in SiOC-1200 and SiOC-1200-H<sub>2</sub>O (Fig. 1(b)). These  $\alpha$ -quartz peaks are absent for the SiOC-1200 and SiOC-1200-H<sub>2</sub>O samples, likely due to the consumption of unresolved Si-OH bonds and  $\alpha$ -quartz by conversion into SiOC through C diffusion. am-SiO<sub>2</sub> and free turbostratic C phases likely reacted at higher pyrolysis temperatures, which was

elucidated through recent ReaxFF/LAMMPS simulation studies.<sup>20,21</sup>

$\alpha$ -Quartz crystallites were postulated to exist predominantly at the interface between MMT nanosheets and the SiOC matrix, as they are persistent in the xMMT-SiOC samples, also consistent with that reported previously.<sup>7</sup> With water vapor-assisted pyrolysis, however,  $\alpha$ -quartz reflections were surprisingly less intense or fully suppressed in 1MMT-900-H<sub>2</sub>O, 2MMT-900-H<sub>2</sub>O, 1MMT-1200-H<sub>2</sub>O, and 2MMT-1200-H<sub>2</sub>O. The intensities of  $\alpha$ -quartz reflections in 5MMT-900 and 5MMT-1200 were identical for both Ar and Ar-20% H<sub>2</sub>O pyrolysis conditions, potentially due to bulk MMT effects from nanosheet coalescence at the high loading content. While all MMT-SiOC-H<sub>2</sub>O samples pyrolyzed at 900 °C and 1200 °C exhibited a more pronounced am-SiO<sub>2</sub> halo, water vapor-assisted pyrolysis suppressed  $\alpha$ -quartz crystallization at the MMT-SiOC interface. This opposed the general notion that water vapor-assisted pyrolysis facilitates SiO<sub>2</sub> formation. Fracture surfaces observed with SEM (Fig. 2(a)) were identical among 2MMT-900 and 2MMT-1200 samples pyrolyzed with either Ar or Ar-20% H<sub>2</sub>O, illustrating dense



**Fig. 1** XRD patterns of xMMT-SiOC (*x* = 0, 2, 5) ceramics pyrolyzed at (a) 900 °C, (b) 1200 °C, and (c) 1400 °C in Ar-20% H<sub>2</sub>O (solid) and Ar (dotted). XRD patterns for 1MMT-SiOC were similar to those of 2MMT-SiOC at all temperatures and were omitted for brevity. The curves were normalized and offset for clarity, and the reference XRD patterns for  $\alpha$ -quartz ( $\alpha$ -Q; 00-033-1161),  $\beta$ -cristobalite ( $\beta$ -Crs; 04-007-2379),  $\beta$ -SiC (04-002-9070), and graphitic carbon (C; 00-041-1487) were provided from ICDD PDF cards.





microstructures with faceted features in MMT-SiOC samples due to fracture at or along MMT domains. No discernible differences in SEM were expected due to similar MMT content across pyrolysis environments and changes to SiO<sub>2</sub> content occurring at the nanoscale. Suppressed  $\alpha$ -quartz crystallization, however, indicated that water vapor-assisted pyrolysis may have reduced overall SiO<sub>2</sub> content and even SiO<sub>2</sub> domain size in MMT-SiOC, contradicting the typical behavior of SiOC with water vapor-assisted pyrolysis. SiOC-900-Ar and SiOC-900-H<sub>2</sub>O as well as SiOC-1200-Ar and SiOC-1200-H<sub>2</sub>O were similarly featureless, as expected.

Further pyrolysis to 1400 °C (Fig. 1(c)) corresponded to the increased presence of am-SiO<sub>2</sub> ( $\sim 22\text{--}24^\circ$  halo), turbostratic C [hexagonal; *P*6<sub>3</sub>/*mmc* (graphite); PDF 00-041-1487] (broad (101) reflection at  $\sim 44^\circ$ ), and poorly crystallized  $\beta$ -SiC [cubic, *F*43*m*; PDF 04-002-9070] (broad reflections at 36°, 60°, and 72° for (111), (220), and (311) planes, respectively). While present in SiOC-1400-H<sub>2</sub>O and 1/2/5MMT-1400-Ar,  $\alpha$ -quartz was not observed in the XRD patterns of 2/5MMT-1400-H<sub>2</sub>O, indicating that water vapor-assisted pyrolysis fully suppressed crystallization of the MMT-SiOC interface. The known increase in SiO<sub>2</sub>

content in SiOC in the presence of water vapor seemingly contradicted the observed inhibition of  $\alpha$ -quartz crystallization, alluding to the reduction of C as the driving mechanism. Free C in SiOC systems originated from excess C-rich functional groups in the preceramic polymer and existed as a continuous network that encapsulated SiO<sub>2</sub> regimes with the SiOC tetrahedra at the interlayer.<sup>22</sup>  $\alpha$ -quartz was theorized to form upon cooling from high-temperature tridymite formed at the MMT-SiOC interface during the pyrolysis.

Identical emergence and strengthening of the 22°  $\beta$ -cristobalite [cubic, *Fd*3*m*; PDF 04-007-2379] reflection among 2/5MMT-1400-Ar and 2/5MMT-1400-H<sub>2</sub>O were attributed to MMT decomposition (an idealized form shown in eqn (2) in which M = Fe, Mg, Al depends on impurities in the clay composition), with the increasing reflection intensity correlating with the MMT content. In addition,  $\beta$ -SiC formation was enhanced in 2/5MMT-1400-H<sub>2</sub>O as indicated by the increased prominence of the (111), (200), and (311) reflections in Fig. 1(c). MMT-SiOC pyrolyzed to 1400 °C in Ar exhibited  $\beta$ -SiC nanowhisker growth (Fig. 2(b–e)) from nanoflake MMT decomposition to cristobalite and mullite, of which the former formed  $\beta$ -SiC *via* carbothermal

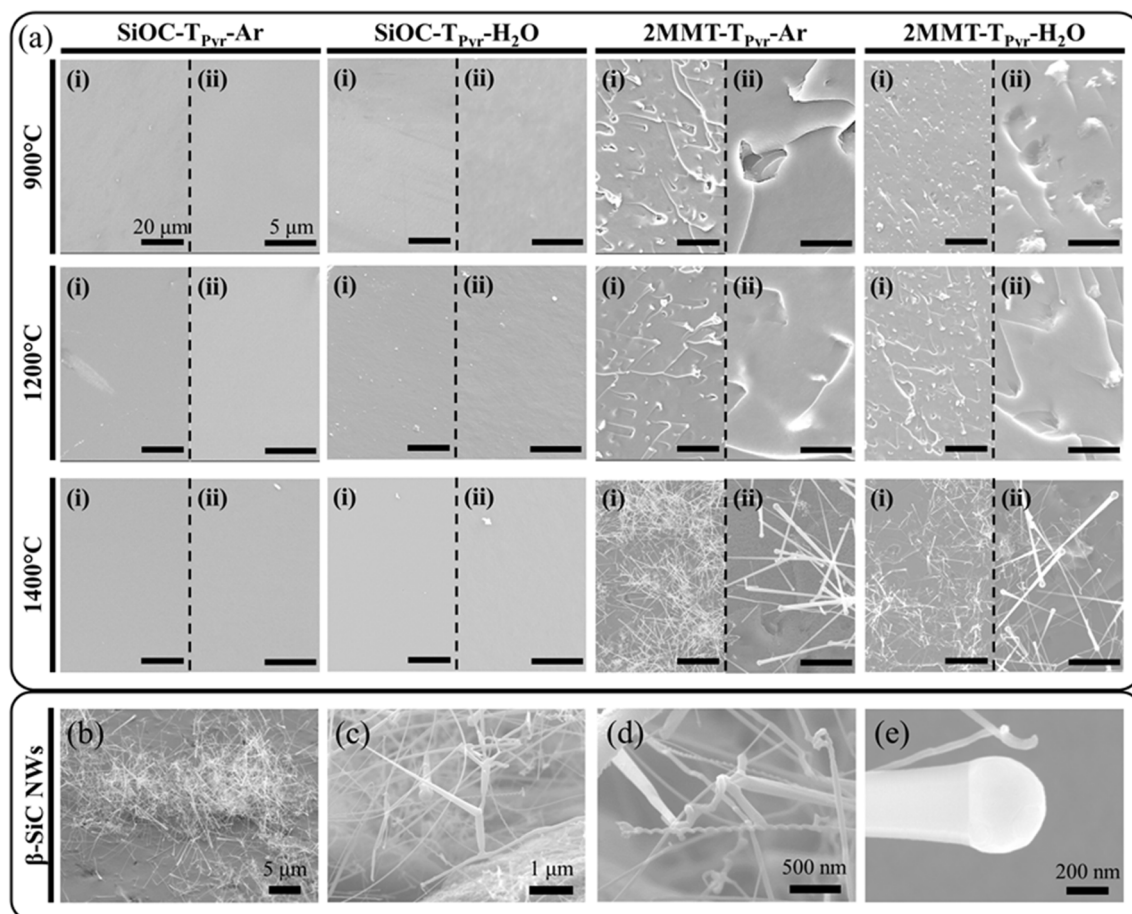
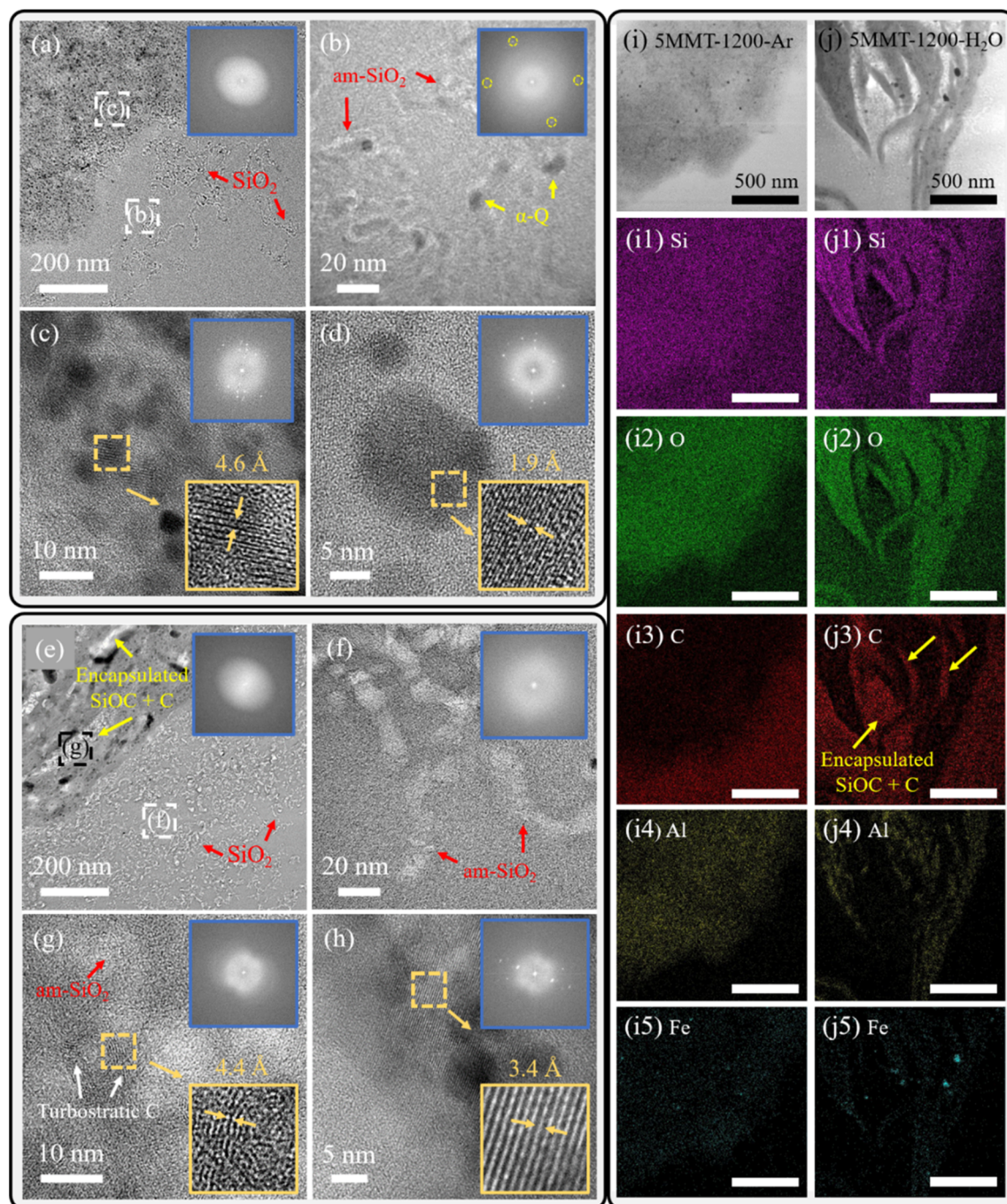


Fig. 2 (a) Fracture surface morphologies of SiOC and 2MMT-SiOC ceramics pyrolyzed to 900 °C, 1200 °C, and 1400 °C with Ar or Ar-20% H<sub>2</sub>O. Scale bars in panels (i) and (ii) in all micrographs are 20 and 5  $\mu$ m, respectively; (b)–(d)  $\beta$ -SiC nanowhisker structures formed during water vapor-assisted pyrolysis of 1/2/5MMT-SiOC to 1400 °C, illustrating a non-steady state and dendritic growth; (e) detached SiO<sub>2</sub> nodule from which  $\beta$ -SiC nanowhiskers form.



reduction with encapsulated and neighboring C regimes (eqn (1)).  $\beta$ -SiC nanowhiskers were abundantly observed on xMMT-1400-Ar and xMMT-1400-H<sub>2</sub>O fracture surfaces, randomly oriented and often protruding from the surface (Fig. 2(b)). SiOC-1400-Ar and SiOC-1400-H<sub>2</sub>O were similarly featureless and lacking  $\beta$ -SiC nanowhiskers, indicating that MMT addition primarily catalyzed nanowhisker growth (Fig. 2(a)). The authors postulated that nanowhisker growth was spurred by the small

amount of mullite from MMT decomposition, because mullite exhibited needle-like morphologies in oxide ceramics<sup>23–25</sup> and *in situ* in a SiC composite.<sup>26</sup> XRD as a bulk technique could not detect mullite, not only because of the studied MMT concentrations being less than 5 wt%, but also the small amount that would theoretically form due to its complex stoichiometry (Al<sub>6</sub>Si<sub>2</sub>O<sub>13</sub>). However, as Al was detected by STEM-EDS in Fig. 3, which will be discussed, mullite was expected to form, although

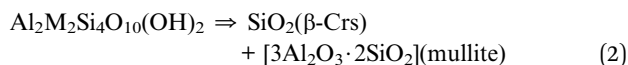
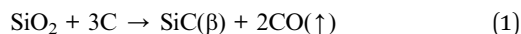


**Fig. 3** (HR)TEM characterization of (a)–(d) 5MMT-1200-Ar and (e)–(h) 5MMT-1200-H<sub>2</sub>O cross-sections prepared by FIB lift-out. Blue-outlined insets show respective FFT patterns and orange-outlined insets show enhanced views of lattice fringes. STEM-EDS elemental maps for Si (1/purple), O (2/green), C (3/red), Al (4/yellow), and Fe (5/blue) of (i) 5MMT-1200-Ar and (j) 5MMT-1200-H<sub>2</sub>O. Scale bars in all (i)/(j) panels are 200 nm.





being undetectable due to the presence of  $\text{Al}_2\text{O}_3$  clusters within a very  $\text{SiO}_2$ -rich environment per the  $\text{SiO}_2$ - $\text{Al}_2\text{O}_3$  phase diagram.<sup>27</sup>

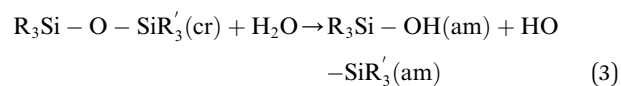


However, these  $\beta$ -SiC nanowhiskers exhibited higher dendritic and non-steady state growth features (Fig. 2(c and d)) compared to those present in MMT-SiOC-Ar.<sup>7</sup> The forked and coiled  $\beta$ -SiC nanowhiskers indicated that there was a general dearth of reactive C species, but this created a stark dichotomy with the XRD patterns showing enhanced  $\beta$ -SiC crystallization in MMT-1400- $\text{H}_2\text{O}$  (Fig. 1(c)). With water vapor-assisted pyrolysis, MMT-1400 ceramics formed more  $\beta$ -SiC, which required more C to be present as  $\beta$ -SiC formation proceeded purely through carbothermal reduction of  $\text{SiO}_2$  in SiOC systems. Non-steady state growth observed at nanowhisker terminations was not attributed to the overall lack of reactive C, but to an insufficient amount relative to the higher concentration of  $\beta$ -SiC nanowhiskers formed near MMT flakes.

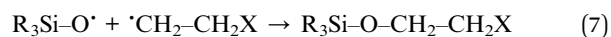
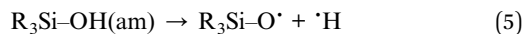
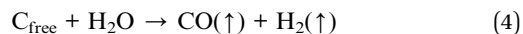
From these observations, we believed that MMT and water vapor-assisted pyrolysis in conjunction both suppressed  $\text{SiO}_2$  crystallization and retained C in MMT-SiOC- $\text{H}_2\text{O}$ , which was further observed and confirmed with (HR)TEM and STEM-EDS nanostructural characterization of 5MMT-1200-Ar and 5MMT-1200- $\text{H}_2\text{O}$  (Fig. 3). The MMT-SiOC interfacial region in 5MMT-1200-Ar (Fig. 3(a)) was peppered with  $\alpha$ -quartz nanoparticles in the generally featureless SiOC matrix. Vein-like  $\text{SiO}_2$  grains were observed to bridge the MMT-SiOC interface, which were further revealed (Fig. 3(b)) to be comprised of both am-SiO<sub>2</sub> (white grains) and  $\alpha$ -quartz (black particles). Crystalline  $\alpha$ -quartz contributions were identified by minor diffraction spots in the FFT inset (yellow circles). MMT was shown previously to catalyze Si-O-Si condensation during pyrolysis of MMT-PSO around 400 °C, a ~50 °C earlier onset than in pure PSO, as the interfacial region between SiOC and MMT formed.<sup>7</sup> This interfacial region was theorized to comprise encapsulated C and SiOC clusters that bridged the coherency difference between crystalline MMT and am-SiO<sub>2</sub>.  $\alpha$ -Quartz nanoparticles on the MMT surface (Fig. 3(c)) and in the interface (Fig. 3(d)) were well-crystallized with lattice fringes measured as 4.6 Å and 1.9 Å and attributed to (100) and (112) planes, respectively.

While 5MMT-1200- $\text{H}_2\text{O}$  exhibited a similar morphology between MMT domains and the SiOC matrix (Fig. 3(e)), the interfacial regions characterized by loss of crystallinity from  $\alpha$ -quartz and encapsulated SiOC/C regimes were observed within or between MMT domains. The MMT-SiOC interfacial region (Fig. 3(f)) exhibited solely am-SiO<sub>2</sub> grains with no crystalline  $\alpha$ -quartz present. Water vapor injection suppressed  $\alpha$ -quartz crystallization through amorphization of crystallized Si-O-Si bonds (eqn (3)), which were shown to exhibit high reactivity with water in *ab initio* molecular dynamic simulations.<sup>28</sup> While high pressures have been shown to spur  $\text{SiO}_2$

amorphization,<sup>29,30</sup> water vapor injection between 500 °C and 700 °C generated  $\cdot\text{H}$  and  $\cdot\text{OH}$  radicals that reacted with strained Si-O-Si bonds. Amorphization of Si-O-Si bonds further relieved lattice strain between crystalline regimes and the amorphous SiOC matrix. In the presence of MMT that created  $\alpha$ -quartz prior to water injection, water vapor promoted am-SiO<sub>2</sub> formation through vitrification of crystalline  $\alpha$ -quartz, accounting for the suppressed reflections observed in XRD (Fig. 1).



MMT regimes effectively shielded C from volatilizing as CO through encapsulation of SiOC/C regions during water vapor injection between 500 °C and 700 °C, as indicated in Fig. 3(e). The lack of such encapsulated regions in 5MMT-1200-Ar indicated that water vapor not only promoted amorphization of Si-O-Si bonds but also C diffusion into  $\text{SiO}_2$ -rich regions. In this regard, water vapor may have led to additional SiOC clusters. HRTEM analysis within encapsulated regions (Fig. 3(g)) revealed turbostratic C diffused between am-SiO<sub>2</sub> grains and characterized by a lattice fringe spacing of 4.4 Å corresponding to (002) C planes. The expanded interlayer spacing was due to C existing as graphene oxide, whose interlayer spacings are affected by O content and have been calculated by molecular dynamics simulations between 3.772 Å and 5.291 Å with increasing O group density from 10% to 50%.<sup>31</sup> Volatilization of C species (e.g.,  $\text{CH}_4$ ,  $\text{H}_2\text{C}=\text{CH}_2$ ,  $\text{H}_3\text{C}-\text{CH}_2=\text{CH}_2$ ) during Ar atmosphere pyrolysis is typical between 400 °C and 700 °C from excess C-rich functional groups in the preceramic polymer,<sup>32-34</sup> and water vapor-assisted pyrolysis further volatilizes free C as CO (eqn (4)).<sup>3,15</sup> However, MMT domains shielded these encapsulated SiOC/C regimes from  $\text{H}_2\text{O}$  oxidation, and C was retained in the matrix. During typical Ar pyrolysis, ethylene and propylene primarily evolved between 500 °C and 700 °C as determined by TGA-MS (thermogravimetric analysis-mass spectroscopy) in related PDC systems.<sup>32,33,35</sup> This specific polysiloxane has not been analyzed with TGA-MS yet but closely resembled polycarbosilane (PCS) systems based on TG curves for PSO-to-SiOC<sup>7</sup> and PCS-to-SiC.<sup>32,33</sup> As water vapor was injected within this temperature range, radical-mediated reactions among Si, O, C, and H led to both C and O incorporation.<sup>34</sup> These reactions are summarized in eqn (5)-(7) in which amorphized Si-OH bonds react with volatilized hydrocarbons (X = H for ethylene and X =  $\text{CH}_3$  for propylene).



EDS elemental maps of 5MMT-1200-Ar (Fig. 3(i)) and 5MMT-1200- $\text{H}_2\text{O}$  (Fig. 3(j)) further revealed encapsulated SiOC/C regimes within MMT domains. MMT domains in both 5MMT-



1200-Ar and 5MMT-1200-H<sub>2</sub>O are Si (Fig. 3(i1)–(j1)) and O (Fig. 3(i2)–(j2)) rich, with C (Fig. 3(i3)–(j3)) generally segregated outside MMT domains. Al (Fig. 3(i4)–(j4)) and Fe (Fig. 3(i5)–(j5)) signatures were detected from the MMT nanosheet. However, as indicated in Fig. 3(j4), C was detected within MMT regimes in 5MMT-1200-H<sub>2</sub>O; this further confirmed that C was retained within MMT regimes and protected from water vapor ingress. Overall, water vapor-assisted pyrolysis produced inverse effects in MMT-SiOC ceramics as MMT domains preferentially retained C and water vapor induced  $\alpha$ -quartz amorphization. Encapsulated SiOC/C regimes further enhanced  $\beta$ -SiC nanowhisker growth in MMT-1400-H<sub>2</sub>O ceramics due to the carbothermal reduction of nanosized SiO<sub>2</sub> and C phases.

The elemental ratios in Table 1, taken in MMT domains and the surrounding SiOC, confirmed that MMT domains were SiO<sub>2</sub>-rich compared to the C-rich SiOC matrix in both 5MMT-1200-Ar and 5MMT-1200-H<sub>2</sub>O. With Ar + H<sub>2</sub>O pyrolysis, Si ( $35.6 \pm 1.2$  at%) and O ( $28.2 \pm 1.3$  at%) contents increased within MMT domains at the expense of C ( $31.3 \pm 1.7$  at%) compared to Si ( $27.4 \pm 1.5$  at%), O ( $20.7 \pm 0.4$  at%), and C ( $47.8 \pm 2.6$  at%) concentrations in 5MMT-1200-Ar. Interestingly, Si ( $\sim 12$  at%), O ( $\sim 16$  at%), and C ( $\sim 71$  at%) ratios for the SiOC matrix were identical in both samples. While the SiOC matrix compositions were similar, the distribution of SiO<sub>x</sub>C<sub>4-x</sub> tetrahedra could differ but such information can only be ascertained with <sup>29</sup>Si NMR.

As previously mentioned, mullite was not observed in XRD or TEM owing to the marginal Al content (2–3 at%) within MMT domains in both samples. Fe was the primary substitutional atom within the nanosheet but was nearly negligible (<2 at%) in both samples; thus, in MMT-SiOC, Fe was not expected to substantially catalyze  $\beta$ -SiC formation as previously reported.<sup>36–41</sup> Encapsulated C regions in 5MMT-1200-H<sub>2</sub>O exhibited a similar composition to the SiOC matrix and reinforced that MMT retained and encapsulated C as both free C and in SiOC during water vapor-assisted pyrolysis.

### 3.2. Physical and mechanical properties

Water vapor-assisted pyrolysis produced MMT-SiOC-H<sub>2</sub>O ceramics that showed a higher retention of C, which had further implications in overall physical properties. Fig. 4 summarizes the ceramic yield (CY, top panel/black) and volumetric shrinkage (VS, bottom panel/blue) of MMT-SiOC-Ar and MMT-

SiOC-H<sub>2</sub>O pyrolyzed at (a) 900 °C, (b) 1200 °C, and (c) 1400 °C. Ceramic residues of MMT-SiOC-H<sub>2</sub>O samples pyrolyzed at 900 °C were 3–5% greater than the corresponding MMT-SiOC-Ar ceramic yields. The ceramic yields of 1/2/5MMT-900-Ar (CYs:  $\sim 77.7 \pm 0.8\%$ ) were not discernible from those of SiOC-900-Ar (CY:  $77.0 \pm 0.8\%$ ) within experimental error. While water vapor-assisted pyrolysis improved the ceramic yield between SiOC-900-Ar and SiOC-900-H<sub>2</sub>O (CY:  $82.0 \pm 0.2\%$ ) by 5%, 1/2MMT-900-H<sub>2</sub>O (CYs:  $\sim 80.5 \pm 0.3\%$ ) and 5MMT-900-H<sub>2</sub>O (CY:  $81.7 \pm 0.5\%$ ) ceramics only showed a 3% improvement compared to 1/2/5MMT-900-Ar. Presumably, MMT inhibited expected reactions between water vapor and the transforming MMT-PSO/MMT-SiOC matrix, hence limiting C removal or OH incorporation caused by water vapor pyrolysis. The observed higher CYs between 2MMT-900-H<sub>2</sub>O and 5MMT-900-H<sub>2</sub>O were attributed to the bulk MMT effects from coalesced MMT regimes.

In addition, water vapor-assisted pyrolysis did not affect volumetric shrinkage (Fig. 4(a)) among SiOC-900-Ar (VS:  $39.9 \pm 0.6\%$ ), SiOC-900-H<sub>2</sub>O (VS:  $40.9 \pm 0.3\%$ ), 1MMT-900-Ar (VS:  $39.4 \pm 0.7\%$ ), and 1MMT-900-H<sub>2</sub>O (VS:  $40.0 \pm 0.4\%$ ). While causing an additional  $\sim 2\%$  shrinkage difference between 2MMT-900-H<sub>2</sub>O (VS:  $41.6 \pm 0.4\%$ ) and 2MMT-900-Ar (VS:  $39.2 \pm 0.9\%$ ), water vapor-assisted pyrolysis accentuated the expected bulk MMT effects at 5% MMT loading with a  $\sim 4\%$  difference between 5MMT-900-H<sub>2</sub>O (VS:  $42.3 \pm 0.2\%$ ) and 5MMT-900-Ar (VS:  $38.4 \pm 0.7\%$ ). Even though it induced notable changes to the pure SiOC matrix, water vapor pyrolysis was not sensitive to increasing MMT content up to 2% as the trends between SiOC-900, 1MMT-900, and 2MMT-900 were identical regardless of the pyrolysis atmosphere. Therefore, water vapor pyrolysis of MMT-SiOC affected SiO<sub>2</sub> phase formation both in the SiOC matrix and in MMT domains, but with inverted effects.

Further pyrolysis to 1200 °C generally contributed to  $\sim 1.5$ – $2\%$  greater ceramic yields (Fig. 4(b)) in SiOC-1200-H<sub>2</sub>O (CY:  $74.5 \pm 0.5\%$ ) and 1/2/5MMT-1200-H<sub>2</sub>O (CY:  $\sim 76.9 \pm 0.6\%$ ) than for Ar-pyrolyzed SiOC-1200 (CY:  $73.5 \pm 0.3\%$ ) and 1/2/5MMT-1200 (CY:  $\sim 75.5 \pm 0.5\%$ ), which were expected from oxygen/hydroxyl incorporation during water vapor-assisted pyrolysis. Ceramic yields among 1/2/5MMT-1200-H<sub>2</sub>O and 1/2/5MMT-1200-Ar were nearly identical and not discernible within experimental error. As shown in Fig. 3(b), the net mass increase corresponded with a general  $\sim 2.5$ – $3\%$  lower volumetric

**Table 1** Elemental ratios from STEM-EDS of Si, O, C, Al, and Fe in 5MMT-1200-Ar and 5MMT-1200-H<sub>2</sub>O. Values represented the averages and standard deviations of 3 different regions within each location: MMT domains (MMT), SiOC matrix (SiOC), and encapsulated C regions in 5MMT-1200-H<sub>2</sub>O (Encaps. C). n.d. – not determined. The EDS spectra are provided in the ESI (Fig. S-5)

Sample	Location	Elemental composition (at%)				
		Si	O	C	Al	Fe
5MMT-1200-Ar	MMT	$27.4 \pm 1.5$	$20.7 \pm 0.4$	$47.8 \pm 2.6$	$2.5 \pm 0.5$	$1.6 \pm 0.2$
	Matrix	$12.1 \pm 0.4$	$16.4 \pm 0.1$	$71.5 \pm 0.3$	n.d.	n.d.
5MMT-1200-H <sub>2</sub> O	MMT	$35.6 \pm 1.2$	$28.2 \pm 1.3$	$31.3 \pm 1.7$	$3.6 \pm 0.1$	$0.6 \pm 0.3$
	Matrix	$12.4 \pm 0.5$	$16.2 \pm 0.1$	$71.3 \pm 0.5$	n.d.	n.d.
	Encaps. C	$15.5 \pm 5.0$	$18.4 \pm 2.0$	$65.4 \pm 7.0$	$0.5 \pm 0.1$	$0.2 \pm 0.1$





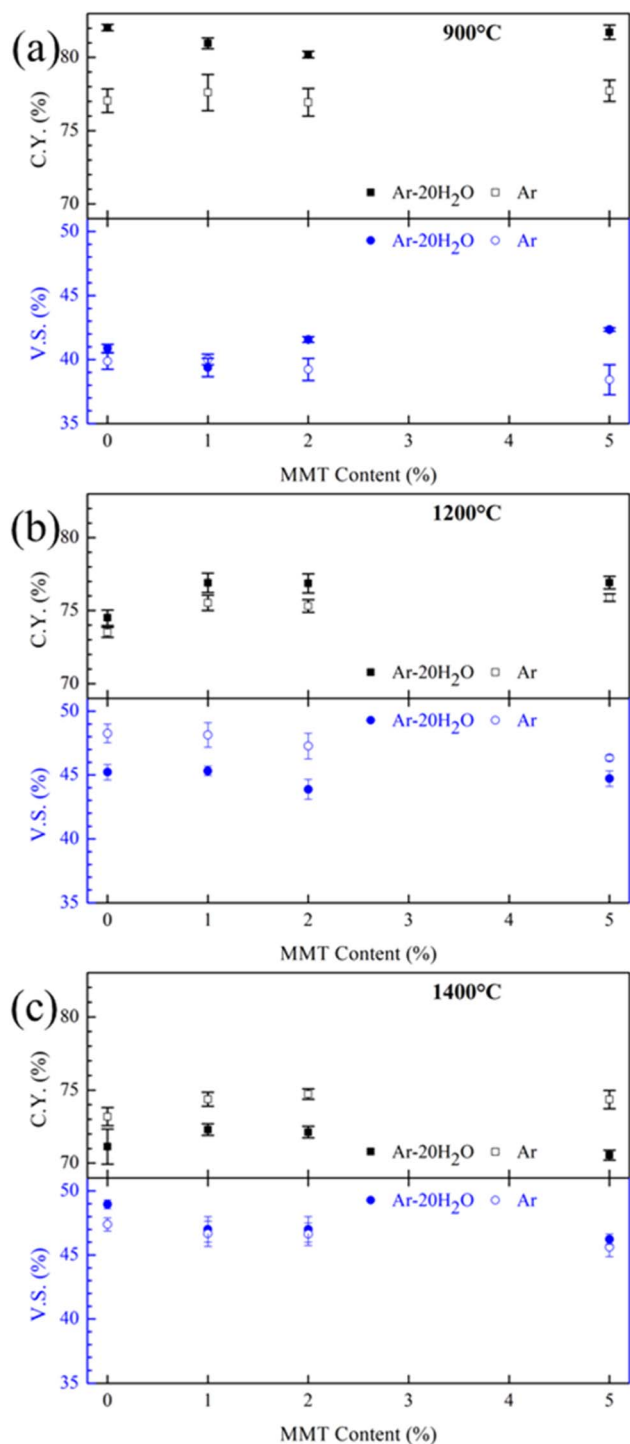


Fig. 4 Ceramic yield (top panel/black) and volumetric shrinkage (bottom panel/blue) of MMT-SiOC ceramics pyrolyzed at (a) 900 °C, (b) 1200 °C, and (c) 1400 °C in either Ar-20% H<sub>2</sub>O (filled symbol) or Ar (unfilled). Error bars indicate a standard deviation of 3–5 samples.

shrinkage across SiOC-1200 and 1/2/5MMT-1200 pyrolyzed with Ar-20% H<sub>2</sub>O compared to the Ar-pyrolyzed analogs.

Volumetric shrinkage in MMT-SiOC-1200 samples was primarily caused by matrix reactions with the water vapor pyrolysis environment: shrinkage increased to  $48.3 \pm 0.6\%$  in SiOC-1200-H<sub>2</sub>O from  $45.2 \pm 0.6\%$  in SiOC-1200-Ar. No

significant differences were observed in 1MMT-1200-H<sub>2</sub>O (VS:  $48.1 \pm 0.8\%$ ) and 1MMT-1200-Ar (VS:  $45.3 \pm 0.4\%$ ) compared to the corresponding SiOC-1200 samples, as 2 wt% MMT was ostensibly the critical loading concentration. The samples with MMT content above 2% exhibited slight VS decreases of  $\sim 1$ – $1.5\%$  across either pyrolysis atmosphere, with 2MMT-1200-H<sub>2</sub>O (VS:  $43.8 \pm 0.8\%$ ) vs. 2MMT-1200-Ar (VS:  $47.3 \pm 0.8\%$ ) and 5MMT-1200-H<sub>2</sub>O (VS:  $44.7 \pm 0.6\%$ ) vs. 5MMT-1200-Ar (VS:  $46.3 \pm 0.5\%$ ), continuing the aforementioned trend between Ar-20% H<sub>2</sub>O and Ar pyrolysis atmospheres. Notably, the increase in ceramic yield among *x*MMT-1200-H<sub>2</sub>O was associated with the absence or suppression of  $\alpha$ -quartz phases (Fig. 1(b)). This indicated that water vapor-assisted pyrolysis of MMT-SiOC did not appreciably introduce SiO<sub>2</sub> around the MMT-SiOC interface but rather predominantly occurred in the SiOC matrix, as suggested earlier. The associated mass gain was believed to originate from retained SiOC or C around the interface that was either resistant to reactions or was introduced *via* water vapor-assisted pyrolysis.

CYs of Ar-pyrolyzed MMT-SiOC decreased by 2–5% between 900 °C and 1200 °C primarily due to H<sub>2</sub> evolution from resolved Si-OH bonds. Transient  $\alpha$ -quartz in SiOC-900 disappeared in SiOC-1200 for both Ar- and water vapor-pyrolyzed ceramics, indicating that 1200 °C pyrolysis was required to achieve a complete SiOC matrix without lingering -OH moieties. This effect was more noticeable in water vapor-pyrolyzed MMT-SiOC with a generally larger CY decrease of 4–8% between 900 °C and 1200 °C because H<sub>2</sub>O injection contributed more -OH and -H groups in the SiOC matrix. While present in 2MMT-900-H<sub>2</sub>O, residual  $\alpha$ -quartz was indiscernible in 2MMT-1200-H<sub>2</sub>O and similarly resolved into am-SiO<sub>2</sub> as in native SiOC.  $\alpha$ -quartz in 5MMT-1200-H<sub>2</sub>O was caused by agglomeration and was inaccessible to H<sub>2</sub>O ingress, while  $\alpha$ -quartz in 2/5MMT-1200-Ar was not reactive with the transforming SiOC matrix between 900 °C and 1200 °C.

The reduction of residual  $\alpha$ -quartz between 900 °C and 1200 °C suggested MMT-SiOC interface refinement as H<sub>2</sub> was released and lingering interfacial Si-OH formed Si-O-Si or Si-O-C bonds. Elastic modulus (*E*) and hardness (*H*) values determined by nanoindentation (the indentation curves are shown in ESI Fig. S-4†) surprisingly revealed differing effects of MMT addition with pyrolysis temperature (Table 2) that were related to this interfacial phenomenon. At 900 °C, mechanical properties decreased with MMT addition from SiOC-900-Ar (*E*:  $58 \pm 1$  GPa; *H*:  $7.1 \pm 0.2$  GPa) to 2MMT-900-Ar (*E*:  $52 \pm 4$  GPa; *H*:  $6.4 \pm 0.3$  GPa); however, at 1200 °C, MMT addition improved the mechanical properties between SiOC-1200-Ar (*E*:  $71 \pm 4$  GPa; *H*:  $8.0 \pm 0.7$  GPa) and 2MMT-1200-Ar (*E*:  $81 \pm 5$  GPa; *H*:  $8.3 \pm 0.6$  GPa). 2 wt% MMT again appeared to be the critical loading amount, below which 1MMT-900-Ar (*E*:  $57 \pm 4$  GPa; *H*:  $6.6 \pm 0.7$  GPa) and 1MMT-1200-Ar (*E*:  $78 \pm 3$  GPa; *H*:  $8.3 \pm 0.5$  GPa) foreshadowed the aforementioned trends observed at 900 °C and 1200 °C. Due to agglomeration effects, mechanical properties were indistinguishable between 2 wt% and 5 wt% MMT contents at both 900 °C (*E*:  $63 \pm 1$  GPa; *H*:  $6.4 \pm 0.2$  GPa) and 1200 °C (*E*:  $58 \pm 1$  GPa; *H*:  $7.1 \pm 0.2$  GPa).



**Table 2** Elastic modulus ( $E$ /GPa) and hardness ( $H$ /GPa) values of MMT–SiOC ceramics pyrolyzed in Ar or Ar–20H<sub>2</sub>O to 900, 1200, and 1400 °C from nanoindentation. The error represented the standard deviation of 20–25 indents

Ceramic sample		$E$ (GPa)			$H$ (GPa)		
		900 °C	1200 °C	1400 °C	900 °C	1200 °C	1400 °C
Ar	SiOC	58 ± 1	71 ± 4	56 ± 4	7.1 ± 0.2	8.0 ± 0.7	6.0 ± 0.7
	1MMT	57 ± 3	78 ± 3	58 ± 5	6.6 ± 0.7	8.5 ± 0.5	6.7 ± 0.7
	2MMT	52 ± 4	81 ± 5	52 ± 4	5.9 ± 0.5	8.3 ± 0.6	5.9 ± 0.7
	5MMT	53 ± 1	77 ± 5	52 ± 5	6.4 ± 0.2	8.6 ± 0.7	6.2 ± 0.8
Ar–20H <sub>2</sub> O	SiOC	—	53 ± 1	59 ± 5	—	5.9 ± 0.2	5.7 ± 0.6
	2MMT	—	59 ± 3	51 ± 5	—	6.4 ± 0.4	5.8 ± 0.6

The elimination of transient  $\alpha$ -quartz and a completely formed amorphous SiOC matrix caused the improvement in Young's modulus between SiOC-900-Ar and SiOC-1200-Ar. At 900 °C, increasing MMT content correlated with increasing concentrations of lingering interfacial Si–OH bonds; that is, as the MMT–SiOC interface was not fully formed, MMT domains were effectively discontinuous inclusions and impeded the mechanical properties in 1/2/5MMT-900-Ar. When the interface was further refined at 1200 °C, MMT strengthened the SiOC matrix even with H<sub>2</sub>O injection due to inherent nanosheet anisotropy. Both  $E$  (53 ± 1 GPa) and  $H$  (5.9 ± 0.2 GPa) expectedly decreased in SiOC-1200-H<sub>2</sub>O because of SiO<sub>2</sub> formation and C removal. MMT domains still abetted the mechanical strength in 2MMT-1200-H<sub>2</sub>O ( $E$ : 59 ± 3 GPa;  $H$ : 6.4 ± 0.2 GPa), although the net reduction was caused by H<sub>2</sub>O ingress into the SiOC matrix. Mechanical properties of xMMT-900-H<sub>2</sub>O were not investigated in light of this interfacial evolution phenomenon and because the  $\alpha$ -quartz content among MMT-900-Ar and MMT-900-H<sub>2</sub>O appeared similar in XRD profiles. While MMT–SiOC interfacial refinement at 1200 °C was assessed through trends among physical and mechanical properties, nanoscale ordering can only be ascertained through advanced TEM techniques (*e.g.*, 4D-STEM) that require a separate study.

Carbothermal reduction at 1400 °C accounted for the general decrease in CY and increase in VS (Fig. 4(c)) through the formation of  $\beta$ -SiC phases and CO evolution in all MMT–SiOC ceramics across either pyrolysis atmosphere. Water vapor-assisted pyrolysis induced an additional ~2–4% mass loss in SiOC-1400-H<sub>2</sub>O (CY: 71.1 ± 0.8%), 1/2MMT-1400-H<sub>2</sub>O (CYs: ~72.2 ± 0.6%), and 5MMT-1400-H<sub>2</sub>O (CY: 70.5 ± 0.4%) compared to SiOC-1400-Ar (CY: 73.2 ± 0.6%) and 1/2/5MMT-1400-Ar (CYs: ~74.5 ± 0.5%). The observed trends were attributed to accelerated CO evolution from  $\beta$ -SiC formation, as the XRD results (Fig. 1(c)) revealed more prominent  $\beta$ -SiC reflections in 1/2/5MMT-1400-H<sub>2</sub>O samples. While water vapor-assisted pyrolysis typically increases SiO<sub>2</sub> and decreases C contents, MMT domains and specifically MMT–SiOC interfaces appeared to be resistant to these effects. Anticipated bulk MMT effects such as nanosheet stacking and increased proximity of MMT regimes from high loading content manifested in a further 2% mass loss in 5MMT-1400-H<sub>2</sub>O. Analogous behavior was not appreciably observed in 5MMT-1400-Ar, presumably due to the abundance of free C.

Volumetric shrinkages among the Ar- and water vapor-pyrolyzed MMT–SiOC-1400 samples were generally identical within experimental error, with a slightly greater shrinkage observed in SiOC-1400-H<sub>2</sub>O (VS: 49.0 ± 0.4%) and SiOC-1400-Ar (VS: 47.4 ± 0.5%) than in MMT-containing ceramics. Among MMT–SiOC-1400 ceramics, pyrolysis atmosphere had little effect on volumetric shrinkage at corresponding MMT content as shown between 1/2MMT-1400-H<sub>2</sub>O (VSs: ~46.6 ± 0.7%) and 1/2MMT-1400-Ar (VSs: ~47.0 ± 0.5%), and between 5MMT-1400-H<sub>2</sub>O (VS: 46.2 ± 0.4%) and 5MMT-1400-Ar (VS: 45.6 ± 0.7%). Above the stable temperature range of SiOC (from 1200–1300 °C), xMMT–SiOC-1400 ceramics illustrated the thermal degradation behavior that would compromise MMT–SiOC microstructures, phase profiles, and physical properties.

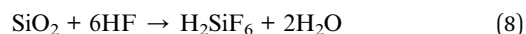
Decreases in  $E$  and  $H$  among 1400 °C ceramics were caused by intrinsic SiOC phase evolution, carbothermal processes (creating inherent porosity), and MMT decomposition. Compared to SiOC-1400-Ar ( $E$ : 56 ± 4 GPa;  $H$ : 6.0 ± 0.7 GPa), MMT decomposition and enhanced  $\beta$ -SiC nanowhisker growth contributed to net reductions observed with increasing MMT content from 1MMT-1400-Ar ( $E$ : 58 ± 5 GPa;  $H$ : 6.7 ± 0.7 GPa) to 2MMT-1400-Ar ( $E$ : 52 ± 4 GPa;  $H$ : 5.9 ± 0.7 GPa) and to 5MMT-1400-Ar ( $E$ : 52 ± 4 GPa;  $H$ : 6.2 ± 0.8 GPa). Strengthening mechanisms imbued by MMT domains at 1200 °C with water vapor were absent in 2MMT-1400-H<sub>2</sub>O ( $E$ : 51 ± 5 GPa;  $H$ : 5.8 ± 0.6 GPa), with a similar reduction compared to SiOC-1400-H<sub>2</sub>O ( $E$ : 59 ± 5 GPa;  $H$ : 5.7 ± 0.6 GPa). MMT–SiOC's mechanical properties across all pyrolysis temperatures resembled those of fused SiO<sub>2</sub> ( $E$ : ~70 GPa) when compared with  $\beta$ -SiC ( $E$ : ~400 GPa) or graphitic C ( $E$ : ~40 GPa (*c*-axis)), as strengthening effectively occurred by diffusion of sp<sup>3</sup>-C into the am-SiO<sub>2</sub>/SiOC matrix.<sup>27,42</sup> The effect of MMT on mechanical properties was self-limited as MMT is intrinsically an oxide and interfacial  $\alpha$ -quartz could weaken the ceramic nanocomposites with an excessively high loading content. Overall, while water vapor-assisted pyrolysis in pure SiOC affected physical and mechanical property changes through reduction of C and incorporation of O, MMT–SiOC ceramics did not show appreciable changes with increasing MMT content. That is, the incorporation of C into MMT domains produced net improvements in CY and VS among all MMT–SiOC–Ar ceramics. This phenomenon also occurred in MMT–SiOC–H<sub>2</sub>O ceramics while C was removed in native SiOC–H<sub>2</sub>O.



### 3.3. Selectively-etched porosity for SiO<sub>2</sub> domain characterization

MMT-SiOC-H<sub>2</sub>O ceramics showed a net reduction of crystallized  $\alpha$ -quartz through XRD and TEM analysis, as water vapor produced sufficient 'OH and 'H radicals to vitrify crystalline Si-O-Si bonds. While water vapor produced a higher concentration of am-SiO<sub>2</sub>, the total concentration and size of these nanosized SiO<sub>2</sub> domains were not apparent from the aforementioned analysis. To characterize SiO<sub>2</sub> content, porous ceramics were prepared by etching SiO<sub>2</sub> domains in MMT-SiOC-Ar and MMT-SiOC-H<sub>2</sub>O with excess HF (eqn (8)), and the effect of water vapor-assisted pyrolysis on selectively-etched porosity and SSAs in MMT-SiOC was determined with N<sub>2</sub> sorption experiments (Fig. 5). In a previous study that mechanistically characterized the role of water vapor pyrolysis and HF etching in phase evolution and porosity, Si 2s (154 eV), Si 2p (103 eV), and O 1s (534 eV) XPS emission intensities were substantially reduced in conjunction with increasing intensity of the C 1s (285 eV) peak in SiOC regardless of Ar or Ar + H<sub>2</sub>O pyrolysis.<sup>3</sup> This phenomenon corresponded to Si and O removal by HF and thus a higher concentration of C in HF-etched SiOC as C-C bonds were

unaffected by HF. These effects were also confirmed by elemental analysis: with Ar pyrolysis, the SiOC composition changed from 28.7 wt% Si to 18.3 wt% Si, 20.3 wt% O to 15.7 wt% O, and 51.1 wt% C to 66.0 wt% C with HF etching; for SiOC prepared with water vapor-assisted pyrolysis, Si and O removal was more apparent with larger decreases in Si and O amounts from 36.3 wt% to 18.1 wt% and from 36.7 wt% to 22.8 wt%, respectively.<sup>3</sup> In addition, XRD showed suppression of amorphous and crystalline SiO<sub>2</sub> reflections after HF etching, and HRTEM confirmed the evolution of 2–5 nm SiO<sub>2</sub> domains and pores before and after HF etching, respectively.<sup>3</sup>



Only 1200 °C and 1400 °C pyrolyzed and etched ceramics were reported in Fig. 5 as XRD patterns, CV/VS trends, and nanoindentation results indicated that H<sub>2</sub> evolution and MMT-SiOC interface formation were not completed at 900 °C. N<sub>2</sub> sorption data for HF-etched 900 °C ceramics are provided in the ESI† as these ceramics had a generally similar trend to HF-etched 1200 °C ceramics. The SSAs of HF-etched MMT-SiOC

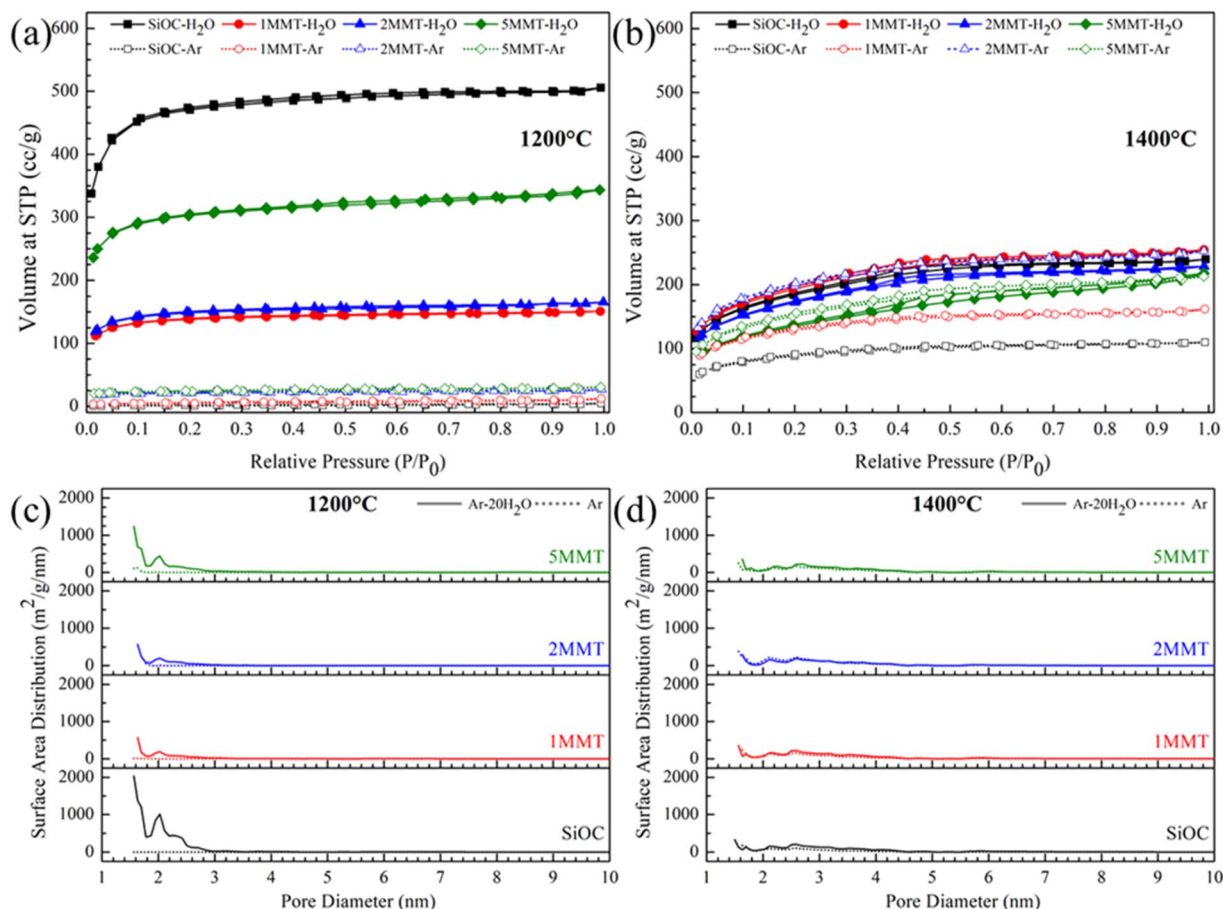


Fig. 5 Porous structure characterization of HF-etched MMT-SiOC-Ar (dashed) and MMT-SiOC-H<sub>2</sub>O (solid) with N<sub>2</sub> sorption. N<sub>2</sub> sorption isotherms of HF-etched xMMT-SiOC (x = 0/black, 1/red, 2/blue, 5/green) pyrolyzed at (a) 1200 °C and (b) 1400 °C. Pore size distributions of HF-etched xMMT-SiOC ceramics pyrolyzed at (c) 1200 °C and (d) 1400 °C as determined by NLDFT fitting. Pore size distributions were offset for clarity.





were reported in Table 3. Selectively etched porosity was representative of the concentration, size, and distribution of SiO<sub>2</sub> domains present in the MMT–SiOC ceramics as Si–C and C–C bonds were resistant to HF.

MMT–SiOC pyrolyzed at 1200 °C in Ar–20% H<sub>2</sub>O exhibited a marked enhancement in pore volumes after HF etching, although porosity was primarily derived through the SiOC matrix. This was shown by marked increases in pore volumes between SiOC-1200-Ar (5 cm<sup>3</sup> g<sup>−1</sup>) and SiOC-1200-H<sub>2</sub>O (505 cm<sup>3</sup> g<sup>−1</sup>). xMMT-1200-Ar were previously shown to marginally yet incrementally enhance pore volumes between 2 and 5 nm after HF etching with 2MMT-1200-Ar and 5MMT-1200-Ar exhibiting the best improvements with pore volumes of 25 cm<sup>3</sup> g<sup>−1</sup> and 30 cm<sup>3</sup> g<sup>−1</sup>, respectively.<sup>7</sup> However, MMT addition contributed to smaller pore volumes in all water vapor-pyrolyzed and HF-etched 1MMT-1200-H<sub>2</sub>O (150 cm<sup>3</sup> g<sup>−1</sup>), 2MMT-1200-H<sub>2</sub>O (165 cm<sup>3</sup> g<sup>−1</sup>), and 5MMT-1200-H<sub>2</sub>O (343 cm<sup>3</sup> g<sup>−1</sup>), indicating that MMT did not promote or accelerate pore formation with water vapor-assisted pyrolysis. The reduction in porosity again contradicted the expected effect of water vapor to grow SiO<sub>2</sub> regimes.

Pore size distributions obtained from NLDFT fitting (Fig. 5(c and d)) revealed a near-unimodal distribution of about 2 nm but less than 3 nm, reflecting the etched SiO<sub>2</sub> domains in SiOC-1200-H<sub>2</sub>O. 1 wt% and 2 wt% MMT content decreased the concentration of these pores before increasing in 5 wt% MMT samples pyrolyzed at 1200 °C, all without altering the intrinsic pore size distribution of SiOC-1200-H<sub>2</sub>O. Therefore, the observed reduction in pore volume in xMMT-1200-H<sub>2</sub>O was not related to SiO<sub>2</sub> growth/coalescence but rather a decrease in SiO<sub>2</sub> content in MMT–SiOC compared to SiOC. SSAs calculated from NLDFT fitting (Table 3) enumerated the decreased SiO<sub>2</sub> content in MMT–SiOC samples at 1200 °C, with SSAs decreasing nearly 75% between SiOC-1200-H<sub>2</sub>O (2081 m<sup>2</sup> g<sup>−1</sup>) and 1MMT-1200-H<sub>2</sub>O (656 m<sup>2</sup> g<sup>−1</sup>). The two-fold increase between 2MMT-1200-H<sub>2</sub>O (707 m<sup>2</sup> g<sup>−1</sup>) and 5MMT-1200-H<sub>2</sub>O (1441 m<sup>2</sup> g<sup>−1</sup>) was associated with additional SiO<sub>2</sub> present in the latter (Fig. 1(b)), presumably due to bulk effects from higher MMT content and excess evolved  $\alpha$ -quartz at the MMT–SiOC interface that could not be incorporated into the amorphous SiOC matrix. Recalling that xMMT-1200-H<sub>2</sub>O exhibited higher ceramic yields than xMMT-1200-Ar, the decrease in SiO<sub>2</sub> concentration was

attributed to the retention of SiOC phases and oxidation resistance of MMT domains in water vapor-pyrolyzed MMT–SiOC.

Pore volumes (Fig. 5(b)) and pore size distributions (Fig. 5(d)) of HF-etched xMMT-1400-Ar and xMMT-1400-H<sub>2</sub>O indicate the dissolution of distinct pore structures of the analogous 1200 °C samples. SiOC-1400-H<sub>2</sub>O (239 cm<sup>3</sup> g<sup>−1</sup>), 1MMT-1400-H<sub>2</sub>O (254 cm<sup>3</sup> g<sup>−1</sup>), 2MMT-1400-H<sub>2</sub>O (228 cm<sup>3</sup> g<sup>−1</sup>), and 5MMT-1400-H<sub>2</sub>O (218 cm<sup>3</sup> g<sup>−1</sup>) all displayed identical pore characteristics. Similarly observed in SiOC-1400-Ar and 1/2/5MMT-1400-Ar, the broadening of pore sizes between 2 and 5 nm illustrated the thermal degradation characteristics and the fate of the SiO<sub>2</sub> phase in MMT–SiOC upon pyrolysis above 1400 °C. As a result, SSAs (Table 3) of SiOC-1400-H<sub>2</sub>O (724 m<sup>2</sup> g<sup>−1</sup>), 1MMT-1400-H<sub>2</sub>O (757 m<sup>2</sup> g<sup>−1</sup>), and 2MMT-1400-H<sub>2</sub>O (657 m<sup>2</sup> g<sup>−1</sup>) were generally identical, with a further reduction in SSA in 5MMT-1400-H<sub>2</sub>O (498 m<sup>2</sup> g<sup>−1</sup>) due to the consumption of SiO<sub>2</sub> during carbo-thermal formation of  $\beta$ -SiC nanowhiskers. At all pyrolysis temperatures, MMT reduced the overall SiO<sub>2</sub> content in MMT–SiOC–H<sub>2</sub>O ceramics compared to native SiOC, and the pore properties were primarily derived from the matrix SiOC, not from MMT regimes or the MMT–SiOC interface. Water vapor-assisted pyrolysis not only suppressed  $\alpha$ -quartz evolution but also decreased the overall SiO<sub>2</sub> content in MMT-1200-H<sub>2</sub>O. The latter surprisingly and severely contradicted typical microstructural evolution behavior in water vapor-assisted SiOC. While an oxide filler, nanostructured MMT did not impart additional SiO<sub>2</sub> domains compared to SiOC in the presence of water vapor, which contradicted the initial hypothesis of this study. Furthermore, the retention or incorporation of C was reinforced as the source of noted improvements in CY upon MMT addition.

### 3.4. Compositional analysis and mechanistic understanding

Water vapor-assisted pyrolysis clearly did not enhance SiO<sub>2</sub> formation in MMT–SiOC, so compositional analysis was performed on SiOC-1200-Ar, SiOC-1200-H<sub>2</sub>O, 2MMT-1200-Ar, and 2MMT-1200-H<sub>2</sub>O to understand the distribution of SiO<sub>x</sub>C<sub>y</sub> and C phases. The concentration of amorphous SiO<sub>x</sub>C<sub>y</sub> (I – SiO<sub>2</sub>, II – SiO<sub>1.5</sub>C<sub>0.25</sub>, III – SiOC<sub>0.5</sub>, IV – SiO<sub>0.5</sub>C<sub>0.75</sub>, V – SiC) phases was quantitatively analyzed with <sup>29</sup>Si NMR as the unique chemical environments about each Si nucleus may be differentiated. Fig. 6(a) depicts the <sup>29</sup>Si NMR spectra of SiOC-1200-Ar (black dotted), SiOC-1200-H<sub>2</sub>O (black solid), 2MMT-1200-Ar (blue dotted), and 2MMT-1200-H<sub>2</sub>O (blue solid) with the expected chemical shifts and Si tetrahedral structures of SiO<sub>2</sub> (−108 ppm), SiO<sub>1.5</sub>C<sub>0.25</sub> (−68 ppm), SiOC<sub>0.5</sub> (−34 ppm), SiO<sub>0.5</sub>C<sub>0.75</sub> (10 ppm), and SiC (−11 ppm) sourced from the literature and indicated.<sup>36,43,44</sup> A small overtone peak (I') at ~−125 ppm causing peak asymmetry of peak I (SiO<sub>2</sub>) was attributed to non-structural 5-coordinated SiO<sub>4</sub> from interactions with neighboring O atoms; while the authors could not find an appropriate reference for this peak in SiOC systems, a similar effect was observed in amorphous Si<sub>3</sub>N<sub>4</sub> systems and the rationale was applied to this system.<sup>45</sup> This I' peak was not included in composition and phase fraction calculations because it did not

**Table 3** Calculated SSAs of etched xMMT–SiOC–Ar and xMMT–SiOC–H<sub>2</sub>O ceramics. SSAs were determined by NLDFT fitting of adsorption branches using a silica adsorbate model with spherical/cylindrical pore structures

Ceramic sample	Specific surface area (m <sup>2</sup> g <sup>−1</sup> )			
	1200 °C		1400 °C	
	Ar	Ar–H <sub>2</sub> O	Ar	Ar–H <sub>2</sub> O
SiOC	3	2081	340	724
1MMT	20	656	499	757
2MMT	105	707	772	657
5MMT	120	1441	574	498





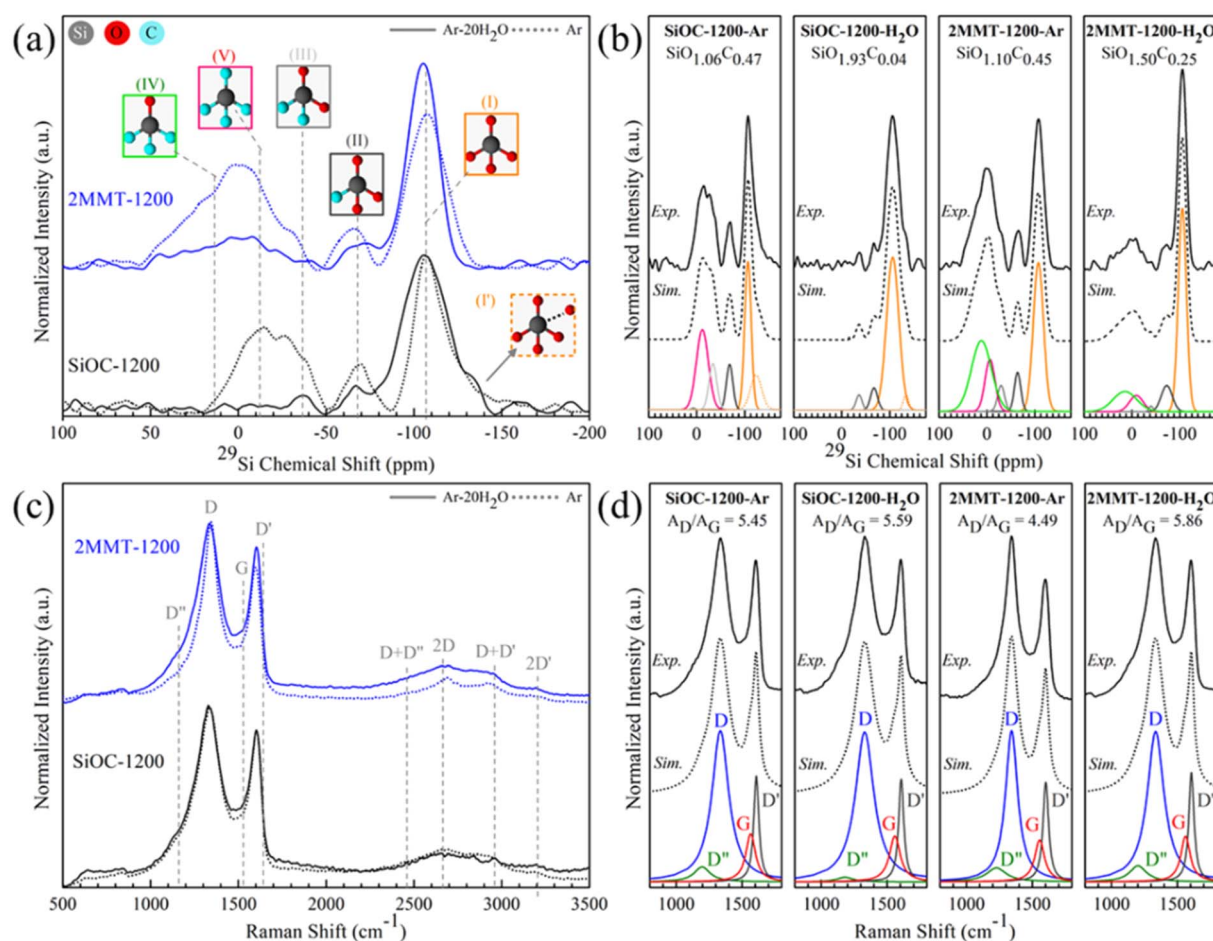


Fig. 6 (a)  $^{29}\text{Si}$  NMR spectra of SiOC-1200-Ar (black/dotted), SiOC-1200- $\text{H}_2\text{O}$  (black/solid), 2MMT-1200-Ar (blue/dotted), and 2MMT-1200- $\text{H}_2\text{O}$  (blue/solid). Peaks I–V indicate the chemical shifts and Si tetrahedral structure of  $\text{SiO}_2$  (–108 ppm/orange),  $\text{SiO}_{1.5}\text{C}_{0.25}$  (–68 ppm/dark grey),  $\text{SiOC}_{0.5}$  (–34 ppm/light gray),  $\text{SiO}_{0.5}\text{C}_{0.75}$  (10 ppm/green), and  $\text{SiC}$  (–11 ppm/pink), respectively; (b) deconvoluted and simulated  $^{29}\text{Si}$  NMR spectra with calculated compositions. Full peak fitting details are provided in the ESI†; (c) Raman spectra of SiOC-1200-Ar (black/dotted), SiOC-1200- $\text{H}_2\text{O}$  (black/solid), 2MMT-1200-Ar (blue/dotted), and 2MMT-1200- $\text{H}_2\text{O}$  (blue/solid); (d) D/G peak deconvolution (D: blue; G: red; D': gray; D'': green) and calculated  $A_D/A_G$  ratio for previous samples. Peak fitting details are provided in the ESI†.

represent a unique Si bonding environment – see the ESI† for details.

SiOC-1200-Ar was revealed to contain  $\text{SiO}_2$ ,  $\text{SiO}_{1.5}\text{C}_{0.25}$ ,  $\text{SiOC}_{0.5}$ , and  $\text{SiC}$  units that converted to predominantly  $\text{SiO}_2$  upon water vapor-assisted pyrolysis in SiOC-1200- $\text{H}_2\text{O}$ . Notably, 2MMT-1200-Ar showed the emergence of  $\text{SiO}_{0.5}\text{C}_{0.75}$  clusters between 50 and 20 ppm, which were not present in SiOC systems and have seldom been reported. MMT generally provides characteristic peaks between –90 and –110 ppm as shown in previous  $^{29}\text{Si}$  NMR characterization studies,<sup>46–48</sup> so the new Si bonding environment in 2MMT-1200-Ar was not from MMT itself but from  $\text{SiO}_{0.5}\text{C}_{0.75}$  clusters presumably at the MMT–SiOC interface. Similar to SiOC-1200- $\text{H}_2\text{O}$ , water vapor-assisted pyrolysis induced restructuring of Si tetrahedra in 2MMT-1200- $\text{H}_2\text{O}$  to predominantly  $\text{SiO}_2$ ; however, some  $\text{SiO}_{0.5}\text{C}_{0.75}$  and  $\text{SiC}$  clusters were retained. While  $\text{SiO}_2$  clusters in SiOC-1200- $\text{H}_2\text{O}$  were completely removed with water vapor-assisted pyrolysis, some  $\text{SiO}_2$  in 2MMT-1200- $\text{H}_2\text{O}$  was oxidation resistant to water vapor ingress between 500 °C and 700 °C,

indicating that these clusters may have formed during the early stages of the polymer-to-ceramic transformation (300 °C to 500 °C).

Gaussian peak deconvolution (Fig. 6(b)) further quantified these trends as the integrated area of each peak provided the relative phase fractions of  $\text{SiO}_x\text{C}_y$  phases and the calculated NMR composition of the SiOC matrix. Full Gaussian peak fitting details may be found in the ESI†. Matrix compositions for SiOC-1200-Ar ( $\text{SiO}_{1.06}\text{C}_{0.47}$ ) and 2MMT-1200-Ar ( $\text{SiO}_{1.10}\text{C}_{0.45}$ ) were nearly identical but were comprised of unique distributions of  $\text{SiO}_x\text{C}_y$  phases. MMT addition primarily induced restructuring of  $\text{SiO}_{1.5}\text{C}_{0.25}$  (11.4% in SiOC-1200-Ar to 6.4% in 2MMT-1200-Ar),  $\text{SiOC}_{0.5}$  (13.8% to 4.7%),  $\text{SiC}$  (32.1% to 13.3%), and  $\text{SiO}_{0.5}\text{C}_{0.75}$  (~0% to 36.5%). The slight increase in  $\text{SiO}_2$  content (37.6% to 39.2%) upon MMT addition was attributed to interfacial  $\alpha$ -quartz and MMT itself. It was postulated that  $\text{SiO}_{0.5}\text{C}_{0.75}$  existed at or near the MMT–SiOC interface, which bridged between crystalline MMT and amorphous SiOC. Furthermore, the higher concentration of retained free C around the MMT



domains contributed to the accelerated  $\beta$ -SiC formation observed in the MMT-SiOC ceramics at 1400 °C (Fig. 3(c)). This would further indicate that MMT was thoroughly incorporated and embedded in the SiOC matrix at the nanoscale as additional  $\text{SiO}_x\text{C}_y$  phases were formed to resolve the incoherency at the interface.

While water vapor-assisted pyrolysis significantly altered the matrix composition ( $\text{SiO}_{1.93}\text{C}_{0.04}$ ) in SiOC-1200- $\text{H}_2\text{O}$  to mainly  $\text{SiO}_2$  (88.4%) with small amounts of  $\text{SiO}_{1.5}\text{C}_{0.25}$  (7.3%) and  $\text{SiOC}_{0.5}$  (4.2%), 2MMT-1200- $\text{H}_2\text{O}$  ( $\text{SiO}_{1.50}\text{C}_{0.25}$ ) was partially resistant to water vapor attack between 500 °C and 700 °C as SiC (8.3%) and  $\text{SiO}_{0.5}\text{C}_{0.75}$  (17.9%) were retained in appreciable amounts and less  $\text{SiO}_2$  (62.3%) was formed. This finding supported the observed decrease in pore volumes and SSAs in xMMT-900- $\text{H}_2\text{O}$  and xMMT-1200- $\text{H}_2\text{O}$  ceramics after HF etching compared to SiOC-900- $\text{H}_2\text{O}$  and SiOC-1200- $\text{H}_2\text{O}$  (Fig. 5). The retained yet decreased SiC and  $\text{SiO}_{0.5}\text{C}_{0.75}$  contents indicated that the domains around MMT or the MMT-SiOC interface were oxidation-resistant, but the size of the MMT-SiOC interface may have decreased upon water vapor pyrolysis. As the SiOC matrix is inherently  $\text{SiO}_2$ -rich as evidenced by STEM-EDS elemental ratios (Table 2) after water vapor pyrolysis, fewer  $\text{SiO}_x\text{C}_y$  clusters were necessary to bridge the coherency gap between crystalline MMT and amorphous SiOC, or am- $\text{SiO}_2$  in the case of SiOC-1200- $\text{H}_2\text{O}$ . The MMT-SiOC interfacial zones effectively decreased in size with Ar +  $\text{H}_2\text{O}$ . However, advanced nanoscale characterization such as synchrotron X-ray or 4D-STEM would be needed to discern changes in crystallinity and amorphous ordering across the MMT-SiOC interface.

As a carbon-rich  $\text{SiO}_x\text{C}_y$  phase,  $\text{SiO}_{0.5}\text{C}_{0.75}$  clusters revealed in  $^{29}\text{Si}$  NMR suggested higher C retention in MMT-SiOC at or near the interface. The nature of C in MMT-SiOC was characterized by Raman spectroscopy (Fig. 6(c)) and elemental analysis (Table 4) to ascertain the effect of MMT on C content and structure. Raman analysis revealed no significant differences in C ordering among SiOC-1200-Ar, SiOC-1200- $\text{H}_2\text{O}$ , 2MMT-1200-Ar, and 2MMT-1200- $\text{H}_2\text{O}$  as C appeared to be comparably disordered as evidenced by the D band at  $\sim 1330\text{ cm}^{-1}$ . The apparent G band at  $\sim 1600\text{ cm}^{-1}$  was determined to have contributions from the actual G band ( $\sim 1550\text{ cm}^{-1}$ ) and an overtone D' band ( $\sim 1620\text{ cm}^{-1}$ ); the presence of the latter was confirmed by D + D' ( $\sim 2950\text{ cm}^{-1}$ ) and 2D' ( $\sim 3200\text{ cm}^{-1}$ ) bands. Superposition of D' and G bands has been commonly reported in disordered carbons, but the difference between the apparent G ( $G_{\text{app}}$ ) and the inferred D' ( $D'_{\text{inf}}$ ) bands from the 2D' band has been shown

to provide more information about the state of C with respect to graphene, graphene oxide (GO), or reduced graphene oxide (rGO).<sup>49,50</sup> Graphene was characterized by a ( $D'_{\text{inf}} - G_{\text{app}}$ ) difference greater than 25, rGO by a difference between 0 and 25, and GO by a difference less than 0.<sup>49</sup> In all samples, 2D' was determined to be  $\sim 3180\text{--}3190\text{ cm}^{-1}$ , yielding  $D'_{\text{inf}}$  of  $\sim 1590\text{--}1595\text{ cm}^{-1}$ , and  $G_{\text{app}}$  was  $\sim 1605\text{ cm}^{-1}$ , indicating that C regimes existed primarily as GO phases with a high degree of disorder. Furthermore, a D + D' overtone band at  $\sim 2450\text{ cm}^{-1}$  revealed a D' band at  $\sim 1180\text{ cm}^{-1}$ , originating from  $\text{sp}^2\text{--sp}^3$  interactions.

Lorentzian peak deconvolution of D and  $G_{\text{app}}$  (Fig. 6(d)) further confirmed the similarity of C regimes among the four ceramic samples with  $A_{\text{D}}/A_{\text{G}}$  ratios of 5.45, 5.59, 4.49, and 5.86 for SiOC-1200-Ar, SiOC-1200- $\text{H}_2\text{O}$ , 2MMT-1200-Ar, and 2MMT-1200- $\text{H}_2\text{O}$ , respectively. Full Lorentzian peak deconvolution and Raman peak information can be found in the ESI.† Due to the significant overlap of D, D', and G bands, the typical  $I_{\text{D}}/I_{\text{G}}$  ratio to determine the degree of disorder was not valid for such disordered carbons, so the integrated area ratio  $A_{\text{D}}/A_{\text{G}}$  was chosen to better quantify D and G contributions.<sup>51,52</sup> The Tuinstra-Koenig equation (eqn (9)) relating lateral crystalline size ( $L_{\alpha}$ ) to  $A_{\text{D}}/A_{\text{G}}$ , with  $\lambda$  the laser excitation wavelength (532 nm), yielded similar  $L_{\alpha}$  values of 3.3, 3.2, 3.9, and 3.1 nm, respectively. The marginal increase in crystallite size at 2MMT-1200-Ar was not deemed significant. Overall, neither water vapor-assisted pyrolysis nor MMT addition altered C structuring in MMT-SiOC.

$$L_{\alpha} = (2.4 \times 10^{-10}) \lambda^4 \left( \frac{A_{\text{D}}}{A_{\text{G}}} \right)^{-1} \quad (9)$$

Elemental analysis (E.A.) was performed to determine the total Si and C content in SiOC-1200-Ar, SiOC-1200- $\text{H}_2\text{O}$ , 2MMT-1200-Ar, and 2MMT-1200- $\text{H}_2\text{O}$ . O was determined as the residual from 100%. Some Si loss was observed during water vapor-assisted pyrolysis between SiOC-1200-Ar (28.5 wt%) and SiOC-1200- $\text{H}_2\text{O}$  (26.3 wt%), as well as between 2MMT-1200-Ar (27.8 wt%) and 2MMT-1200- $\text{H}_2\text{O}$  (26.3 wt%). While this was attributed to a small amount of Si evolution during water vapor injection, the overall Si content was generally invariant. Agreeing with previous studies,<sup>3,15,16</sup> water vapor pyrolysis aggressively removed C between SiOC-1200-Ar (31.7 wt%) and SiOC-1200- $\text{H}_2\text{O}$  (20.6 wt%) through oxidation of free C. However, not only did MMT addition increase the C content by

**Table 4** Elemental analysis (E.A.), NMR, and calculated  $\text{SiO}_x\text{C}_y$  +  $\text{zC}_{\text{free}}$  compositions of SiOC-1200-Ar, SiOC-1200- $\text{H}_2\text{O}$ , 2MMT-1200-Ar, and 2MMT-1200- $\text{H}_2\text{O}$ . As the NMR composition only considered C and O bonded to Si, the free C phase content and composition were extracted from the E.A. composition

Sample	Elemental analysis (E.A.)			E.A. Composition	NMR composition	$\text{SiO}_x\text{C}_y$ + $\text{zC}_{\text{free}}$
	Si (wt%)	O (wt%)	C (wt%)			
SiOC-1200-Ar	28.5	39.8	31.7	$\text{SiO}_{2.45}\text{C}_{2.59}$	$\text{SiO}_{1.06}\text{C}_{0.47}$	$\text{SiO}_{1.06}\text{C}_{0.47}$ + $2.1\text{CO}_{0.65}$
SiOC-1200- $\text{H}_2\text{O}$	26.3	53.2	20.6	$\text{SiO}_{3.54}\text{C}_{1.83}$	$\text{SiO}_{1.93}\text{C}_{0.04}$	$\text{SiO}_{1.93}\text{C}_{0.04}$ + $1.8\text{CO}_{0.90}$
2MMT-1200-Ar	27.8	25.7	46.6	$\text{SiO}_{1.62}\text{C}_{3.92}$	$\text{SiO}_{1.10}\text{C}_{0.45}$	$\text{SiO}_{1.10}\text{C}_{0.45}$ + $3.5\text{CO}_{0.15}$
2MMT-1200- $\text{H}_2\text{O}$	26.3	29.8	43.9	$\text{SiO}_{1.98}\text{C}_{3.90}$	$\text{SiO}_{1.50}\text{C}_{0.25}$	$\text{SiO}_{1.50}\text{C}_{0.25}$ + $3.7\text{CO}_{0.13}$



15 wt% in 2MMT-1200-Ar (46.6 wt%), but also MMT addition proved to show resistance to water vapor ingress as the C content in 2MMT-1200-H<sub>2</sub>O (43.9 wt%) did not decrease similarly to that in SiOC-1200-H<sub>2</sub>O.

From the overall (E.A.) composition calculated from Si, O, and C content (Table 4) and the matrix composition calculated from <sup>29</sup>Si NMR (Fig. 6(c) and also provided in Table 4), the stoichiometric SiO<sub>x</sub>C<sub>y</sub> and free C phase compositions were described as SiO<sub>x</sub>C<sub>y</sub> + zC<sub>free</sub>. In each sample, the O coefficient from E.A. was higher than the matrix O content from <sup>29</sup>Si NMR, so the excess O was assigned to the free C phase in the form of GO following Raman analysis. While the relative free C to SiO<sub>x</sub>C<sub>y</sub> ratio in SiOC-1200-Ar (SiO<sub>1.06</sub>C<sub>0.47</sub> + 2.1 CO<sub>0.65</sub>) did not appreciably decrease in SiOC-1200-H<sub>2</sub>O (SiO<sub>1.93</sub>C<sub>0.04</sub> + 1.8 CO<sub>0.90</sub>), the free C phase was more oxidized, as was expected from prior studies. Not only did the free C phase content increase with MMT addition for both 2MMT-1200-Ar (SiO<sub>1.10</sub>C<sub>0.45</sub> + 3.5 CO<sub>0.15</sub>) and 2MMT-1200-H<sub>2</sub>O (SiO<sub>1.50</sub>C<sub>0.25</sub> + 3.7 CO<sub>0.13</sub>), but also the phase was less oxidized. Overall, while MMT addition did not alter the size or bonding characteristics of the free C phase, it encapsulated more C within the MMT-SiOC interface that was resistant to water vapor oxidation during the polymer-to-ceramic transformation. Even if potentially further oxidized, C was preferentially retained in the MMT-SiOC matrix as GO and not evolved as CO.

Fig. 7 summarizes the phase evolution mechanisms in MMT-SiOC-Ar and MMT-SiOC-H<sub>2</sub>O ceramics from the pre-ceramic MMT-PSO through stages before and after water vapor injection between 500 °C and 1200 °C. During phase evolution in the polymer-to-ceramic transformation, MMT regimes retained more C from the preceramic polymer within the interfacial region between 400 °C and 500 °C when silane condensation reactions occurred earlier with MMT addition, as shown by TGA.<sup>7,35</sup> In these regions of nascent SiOC, the matrix retained more C around MMT regimes acting as a SiO<sub>2</sub> source to maintain the overall matrix composition of the MMT-SiOC ceramics, which were shown earlier by <sup>29</sup>Si NMR to be identical between SiOC-1200-Ar and 2MMT-1200-Ar. As this event occurred prior to water vapor injection in the 2MMT-SiOC

samples, nascent SiOC regimes about MMT were not as susceptible to oxidation as similar phases forming in SiOC around 475–500 °C.<sup>7</sup> While producing SiO<sub>2</sub> structures and removing C in the pure SiOC matrix, water vapor-assisted pyrolysis preferentially incorporated and retained C within amorphized α-quartz domains. In MMT-SiOC-Ar, α-quartz exists solely within the MMT interfacial region with a small concentration of SiO<sub>2</sub> clusters forming in the matrix. In the presence of H<sub>2</sub>O, there is a higher concentration of am-SiO<sub>2</sub> in the SiOC matrix but α-quartz is transformed to am-SiO<sub>2</sub> with more C retained within the MMT-SiOC interface.

In SiOC systems, free C is modeled as surrounding and diffusing into SiO<sub>2</sub> clusters to form SiO<sub>x</sub>C<sub>y</sub> tetrahedra.<sup>22</sup> The emergence of the SiO<sub>0.5</sub>C<sub>0.75</sub> phase in 2MMT-1200-Ar supported this mechanistic explanation. In MMT-SiOC, additional C was retained to resolve structural incoherency between crystalline MMT and amorphous SiOC. Coherency and interfacial bonding occurred through SiO<sub>0.5</sub>C<sub>0.75</sub>, SiC, and SiOC<sub>0.5</sub> phases, which formed through C diffusion into encapsulated regimes. Forming earlier in MMT-SiOC than in pure SiOC, these regimes were resistant to water vapor attack. The effects of water vapor on phase formation, composition, and selectively etched porosity were then only a function of the SiOC matrix with MMT decreasing the total SiO<sub>2</sub> content.

The presence of MMT appreciably altered the effects of Ar + H<sub>2</sub>O pyrolysis, in which SiO<sub>2</sub> suppression and C retention starkly contradicted expected results from previous studies. Water vapor-assisted pyrolysis further elucidated that the MMT-SiOC interface was active in radical-mediated reactions during the polymer-to-ceramic transformation, essentially classifying 2D nanostructured MMT as a hybrid active-passive filler for PDCs. The findings of this study further MMT's potential as a sustainable, naturally derived reinforcement in high-temperature ceramic composites through a fundamental understanding of MMT's contribution to phase evolution and material properties in the MMT-SiOC system. In particular, this methodology could be transferred to other 2D ceramic nanocomposite systems: for example, Ti<sub>3</sub>C<sub>2</sub>T<sub>x</sub> (MXene)-SiOC nanocomposites, which are attractive candidate materials for electrically-conductive yet thermally-insulating applications, prepared in a similar manner, were also characterized with interfacial α-quartz,<sup>53,54</sup> whose suppression can expand the viable temperature range above ~600 °C as potential fracture from the α-to-β transformation was mitigated.

## 4. Conclusions

In this study, MMT-SiOC-Ar and MMT-SiOC-H<sub>2</sub>O nanocomposites were prepared to mechanistically determine the effect of water vapor on SiO<sub>2</sub> phase evolution and material properties. Water vapor-assisted pyrolysis typically removes free C and drastically increases amorphous SiO<sub>2</sub> clusters between 2 nm and 5 nm in pure SiOC. MMT domains not only inhibited H<sub>2</sub>O ingress to reduce SiO<sub>2</sub> content compared to native SiOC (as shown by N<sub>2</sub> sorption experiments) but also selectively amorphized crystalline α-quartz and encapsulated/shielded C from volatilization (as confirmed by (HR)TEM analysis). H<sub>2</sub>O

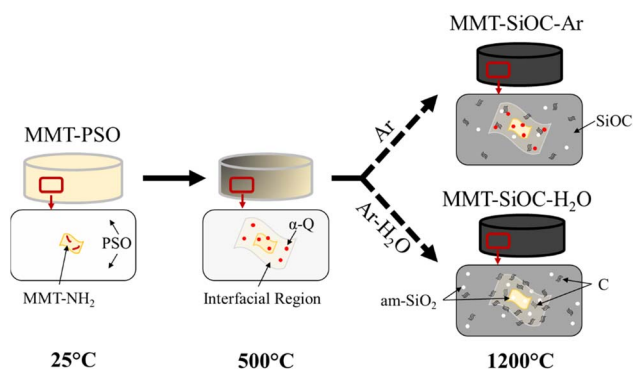


Fig. 7 Schematic of phase evolution mechanisms in MMT-SiOC-Ar and MMT-SiOC-H<sub>2</sub>O ceramics. With water vapor-assisted pyrolysis, α-quartz (α-Q) converted to am-SiO<sub>2</sub>, and more free C was retained within the MMT-SiOC interfacial regime.





injection activated and vitrified Si–O–Si bonds, which then reacted with hydrocarbon gases during pyrolysis to retain C.  $^{29}\text{Si}$  NMR characterization revealed the emergence of  $\text{SiO}_{0.5}\text{C}_{0.75}$  clusters in MMT–SiOC that were postulated to form within the MMT–SiOC interface and were retained after Ar– $\text{H}_2\text{O}$  pyrolysis. Encapsulated SiOC/C regimes further contributed to enhanced  $\beta$ -SiC nanowhisker growth as shown by SEM and XRD. This study challenged conventional assumptions about reactive pyrolysis in PDCs through new findings of encapsulation, shielding, and retention of C as graphene oxide in MMT–SiOC– $\text{H}_2\text{O}$  ceramics under the simultaneous presence of MMT and water vapor and a mechanistic understanding of interfacial phenomena and phase formation in MMT–SiOC ceramic nanocomposites. It opens new avenues for tailoring nanocomposite interfaces in designing multifunctional or porous ceramics with targeted phase compositions and using natural nanostructured additives to achieve high-performance materials through sustainable methods.

## Data availability

The raw and processed data required to reproduce these findings are available upon request.

## Author contributions

Advaith V. Rau: conceptualization, methodology, validation, formal analysis, investigation, resources, data curation, writing – original draft, writing – review & editing, supervision, visualization. Kathy Lu: resources, writing – review & editing, supervision, project administration, funding acquisition.

## Conflicts of interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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## References

- 1 K. Lu, *Mater. Sci. Eng. R Rep.*, 2015, **97**, 23–49.
- 2 K. Lu and D. Erb, *Int. Mater. Rev.*, 2017, **63**, 139–161.
- 3 K. Lu and J. Li, *J. Eur. Ceram. Soc.*, 2016, **36**, 411–422.
- 4 D. Erb and K. Lu, *J. Am. Ceram. Soc.*, 2018, **101**, 5378–5388.
- 5 T. Chen, Y. Yuan, Y. Zhao, F. Rao and S. Song, *Langmuir*, 2019, **35**, 2368–2374.
- 6 E. Bernardo, P. Colombo and E. Manias, *Ceram. Int.*, 2006, **32**, 679–686.
- 7 A. V. Rau and K. Lu, *Small*, 2025, **21**, 2408218.
- 8 T. Liang, Y.-L. Li, D. Su and H.-B. Du, *J. Eur. Ceram. Soc.*, 2010, **30**, 2677–2682.
- 9 G. S. Bibbo, P. M. Benson and C. G. Pantano, *J. Mater. Sci.*, 1991, **26**, 5075–5080.
- 10 P. Colombo, T. E. Paulson and C. G. Pantano, *J. Sol-Gel Sci. Technol.*, 1994, **2**, 601–604.
- 11 H. Ding and M. J. Demkowicz, *Sci. Rep.*, 2015, **5**, 13051.
- 12 M. Narisawa, H. Hokazono, K. Mitsuhara, H. Inoue and T. Ohta, *J. Ceram. Soc. Jpn.*, 2016, **124**, 1094–1099.
- 13 G. D. Sorarù, C. Tavonatti, L. Kundanati, N. Pugno and M. Biesuz, *J. Am. Ceram. Soc.*, 2020, **103**, 6519–6530.
- 14 A. H. Tavakoli, M. M. Armentrout, M. Narisawa, S. Sen, A. Navrotsky and R. Riedel, *J. Am. Ceram. Soc.*, 2015, **98**, 242–246.
- 15 J. Li and K. Lu, *J. Am. Ceram. Soc.*, 2015, **98**, 2357–2365.
- 16 K. Bawane, D. Erb and K. Lu, *J. Eur. Ceram. Soc.*, 2019, **39**, 2846–2854.
- 17 N. Yang, M. Gao, J. Li and K. Lu, *J. Am. Ceram. Soc.*, 2019, **103**, 145–157.
- 18 E. Ionescu, C. Terzioglu, C. Linck, J. Kaspar, A. Navrotsky, R. Riedel and L. Gauckler, *J. Am. Ceram. Soc.*, 2013, **96**, 1899–1903.
- 19 M. Narisawa, K. Yamada, U. Sakura and H. Inoue, *J. Am. Ceram. Soc.*, 2019, **102**, 6449–6454.
- 20 H. Chaney and K. Lu, *Ceram. Int.*, 2023, **49**, 10193–10197.
- 21 K. Lu and H. Chaney, *J. Am. Ceram. Soc.*, 2023, **106**, 2737–2743.
- 22 A. Saha and R. Raj, *J. Am. Ceram. Soc.*, 2007, **90**, 578–583.
- 23 P. V. López, A. Mocciano, M. F. Hernández, D. Richard and N. M. Rendtorff, *Int. J. Appl. Ceram. Technol.*, 2024, **21**, 4081–4090.
- 24 M. Awaad, S. H. Kenawy and S. M. Naga, *The International Conference on Applied Mechanics and Mechanical Engineering*, 2008, **vol. 13**, 37–49.
- 25 M. Romero, I. Padilla, M. Contreras and A. López-Delgado, *Minerals*, 2021, **11**, 332.
- 26 M. Wang, L. Li, Z. Li, M. Du, J. Liu, Q. Zhou and Z. Feng, *Ceram. Int.*, 2019, **45**, 11131–11135.
- 27 *Springer Handbook of Glass*, Springer, Cham, 1 edn, 2019.
- 28 G. V. Gibbs, D. F. Cox, M. B. Boisen, R. T. Downs and N. L. Ross, *Phys. Chem. Miner.*, 2003, **30**, 305–316.
- 29 R. M. Hazen, L. W. Finger, R. J. Hemley and H. K. Mao, *Solid State Commun.*, 1989, **72**, 507–511.
- 30 Y. Nakamura, J. Muto, H. Nagahama, I. Shimizu, T. Miura and I. Arakawa, *Geophys. Res. Lett.*, 2012, **39**, L21303.
- 31 V. Proust, M. C. Bechelany, R. Ghisleni, M.-F. Beaufort, P. Miele and S. Bernard, *J. Eur. Ceram. Soc.*, 2016, **36**, 3671–3679.
- 32 R. Sreeja, B. Swaminathan, A. Painuly, T. V. Sebastian and S. Packirisamy, *Mater. Sci. Eng., B*, 2010, **168**, 204–207.
- 33 J. L. Plawsky, F. Wang and W. N. Gill, *AIChE J.*, 2002, **48**, 2315–2323.





- 34 C. Garcia-Garrido, P. E. Sanchez-Jimenez, L. A. Perez-Maqueda, A. Perejon and J. M. Criado, *Phys. Chem. Chem. Phys.*, 2016, **18**, 29348–29360.
- 35 A. V. Rau, K. Knott and K. Lu, *Mater. Chem. Front.*, 2021, **5**, 6530–6545.
- 36 B. Du, J. Qian, P. Hu, C. He, M. Cai, X. Wang and A. Shui, *J. Am. Ceram. Soc.*, 2019, **103**, 1732–1743.
- 37 Y. Gao, Y. Bando, T. Sato and Y. Kitami, *Jpn. J. Appl. Phys.*, 2001, **40**, 1065–1067.
- 38 Y. Gou, X. Tong, Q. Zhang, B. Wang, Q. Shi, H. Wang, Z. Xie and Y. Wang, *Ceram. Int.*, 2016, **42**, 681–689.
- 39 Z. Yu, L. Yang, H. Min, P. Zhang, C. Zhou and R. Riedel, *J. Mater. Chem. C*, 2014, **2**, 1057–1067.
- 40 M. Hojamberdiev, R. M. Prasad, C. Fasel, R. Riedel and E. Ionescu, *J. Eur. Ceram. Soc.*, 2013, **33**, 2465–2472.
- 41 A. Jain, S. P. Ong, G. Hautier, W. Chen, W. D. Richards, S. Dacek, S. Cholia, D. Gunter, D. Skinner, G. Ceder and K. A. Persson, *APL Mater.*, 2013, **1**, 011002.
- 42 C. Stabler, E. Ionescu, M. Graczyk-Zajac, I. Gonzalo-Juan and R. Riedel, *J. Am. Ceram. Soc.*, 2018, **101**, 4817–4856.
- 43 I. Hung, E. Ionescu, J. Sen, Z. Gan and S. Sen, *J. Phys. Chem. C*, 2021, **125**, 4777–4784.
- 44 G. Mera, A. Navrotsky, S. Sen, H.-J. Kleebe and R. Riedel, *J. Mater. Chem. A*, 2013, **1**, 3826.
- 45 I. Ponomarev and P. Kroll, *Materials*, 2018, **11**, 1646.
- 46 I. W. M. Brown, K. J. D. MacKenzie and R. H. Meinhold, *J. Mater. Sci.*, 1987, **22**, 3265–3275.
- 47 J. Grandjean, J. Bujdák and P. Komadel, *Clay Miner.*, 2018, **38**, 367–373.
- 48 A. K. Mishra, S. Allauddin, R. Narayan, T. M. Aminabhavi and K. V. S. N. Raju, *Ceram. Int.*, 2012, **38**, 929–934.
- 49 A. A. King, B. R. Davies, N. Noorbehesht, P. Newman, T. L. Church, A. T. Harris, J. M. Razal and A. I. Minett, *Sci. Rep.*, 2016, **6**, 19491.
- 50 Z. Li, L. Deng, I. A. Kinloch and R. J. Young, *Prog. Mater. Sci.*, 2023, **135**, 101089.
- 51 L. G. Cançado, M. G. da Silva, F. H. E. Martins, H. Ferreira, K. Kampioti, K. Huang, A. Pénicaud, C. A. Achete, R. B. Capaz and A. Jorio, *2D Materials*, 2017, **4**, 025039.
- 52 N. Yang and K. Lu, *Carbon*, 2021, **171**, 88–95.
- 53 A. V. Rau, Structural Characterization and Material Property Evaluation in Polymer Derived SiOC Ceramics and Ceramic Nanocomposites, *PhD thesis*, Virginia Polytechnic Institute and State University, 2024, pp. 120–144.
- 54 A. V. Rau and K. Lu, *Adv. Compos. Hybrid Mater.*, 2025, **8**, 253.

