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Development of calcium stabilized nitrogen rich alpha-sialon ceramics along the Si3N4-1/2Ca3N2:3AlN line using spark plasma sintering

Bilal Anjum Ahmed
National Kaohsiung University of Science and Technology

Abbas Saeed Hakeem
King Fahd University of Petroleum & Minerals

Tahar Laoui (tlaoui@hotmail.com)
University of Sharjah https://orcid.org/0000-0002-9527-6610

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Abstract

Calcium stabilized nitrogen rich sialon ceramics having a general formula of $Ca_xSi_{12-2x}Al_{2x}N_{16}$ with x value in the range of 0.2-2.2 for compositions lying along the Si_3N_4-1/2Ca_3N_2:3AlN line were synthesized using nano/submicron size starting powder precursors and spark plasma sintering (SPS) technique. The development of calcium stabilized nitrogen rich sialon ceramics at a significantly low sintering temperature of 1500°C (typically reported temperature of 1700°C or greater) remains to be the highlight of the present study. The SPS processed sialons were characterized for their microstructure, phase and compositional analysis, physical and mechanical properties. Furthermore, a correlation was developed between the lattice parameters and the content (x) of alkaline metal cation in the alpha-sialon phase. Well densified single-phase nitrogen rich alpha-sialon ceramics were achieved in the range of $0.4 < x < 1.6$. A nitrogen rich alpha-sialon sample possessing a maximum hardness of 22.4 GPa and fracture toughness of 6.1 MPa.m$^{1/2}$ was developed.

1. Introduction

Sialon materials is a class of ceramics that have been thoroughly studied with respect to their phase stability regimes, thermo-mechanical properties, photoluminescence and oxidation behavior [1–11]. Most of the work on sialon ceramics has been concentrated around the two commonly known phases, namely alpha-sialon and beta-sialon. As an alloy of silicon nitride, beta-sialon having a general formula of $Si_{6-z}Al_zO_zN_{8-z}$, is formed as a result of chemical replacement of silicon-nitrogen bonds with aluminum-oxygen bonds[12].

Compositions belonging to alpha-sialon phase are generally defined by $M_x^vSi_{12-(m+n)}Al_{m+n}O_nN_{16-n}$ (x and v represent solubility and valence of the cation M in alpha-sialon structure, respectively) where $x < 2$, $x = mv$, and m (Al-N) and n (Al-O) replace (m+n) (Si-N) bonds[13]. The substitution of silicon-nitrogen bonds with aluminum-nitrogen and aluminum-oxygen bonds results in a charge imbalance which is neutralized by the incorporation of M^{v+} ions such as Li^+, Ca^{2+}, Mg^{2+}, Y^{3+} and lanthanide ions[14–17]. Studies have been reported on synthesis and phase stability regime of single phase alpha sialons [14–19]. Several compositions of yttrium-stabilized alpha-sialons along the nitrogen rich line of Si_3N_4–YN:3AlN were synthesized by Sun et al where the solubility limit of yttrium cation in alpha sialon was found out to be $0.43<x<0.8$ [1]. Zhen-Kun et al. studied the formation of oxygen rich yttrium stabilized alpha-sialons along the Si_3N_4–Y_2O_3:9AlN and the solubility limit of yttrium cations was reported as $0.33<x<0.67$ [14]. It is well known that synthesis of alpha-sialons occurs via solution-reprecipitation mechanism, which involves the formation of an intermediary oxy-nitride liquid phase as a result of the reaction between the starting oxide and nitride precursors [20,21].

In contrast to alpha-sialons synthesized using yttrium or rare earth stabilizing cations, calcium stabilized alpha-sialons have gained considerable attention due to the higher solubility of calcium cation as compared to the former (maximum solubility value x_{max} of 1.6 for Ca^{+2} as compared to x_{max} value of 1.0

for Yb^{+2}) [15,17]. Furthermore, Hassan Mandal, in his study on post sintering heat treatment of alpha sialons, reported that in contrast to rare-earth stabilized alpha-sialons, calcium stabilized alpha-sialons depicted complete resistance towards alpha to beta phase transformation in the temperature range of 1450-1500°C [22].

The higher stability of calcium stabilized alpha-sialons has been attributed to the lower valency of calcium cation, since the high temperature stability of alpha-sialon increases with increase in the solubility of charge stabilizing cation (x), which in turn increases with decrease in cation valency [23].

Figure 1 shows the Janecke prism and the calcium alpha-sialon plane highlighting the single-phase alpha-sialon region (oxygen rich as well as nitrogen rich) along with the neighboring phases at 1800°C [24]. Several studies conducted on the synthesis of oxygen rich alpha-sialons have reported formation of elongated morphology of alpha-sialon as compared to the generally known equiaxed morphology, thus resulting in considerable improvement in the toughness of these materials [25–27]. However, there are very few studies that have been reported on the synthesis of nitrogen-rich alpha-sialons, mainly due to the fact that they are difficult to densify [23,28]. Studies on the synthesis of calcium stabilized nitrogen rich alpha-sialons using all nitride precursors is even more scarce in the literature due to the difficulty in handling and storage of calcium nitride [23].

Xie.et al worked on the synthesis of calcium stabilized nitrogen rich alpha-sialons along the Si_3N_4-1/2Ca_3N_2:3AlN line at 1700°C using hot pressing technique [23]. The limits of single phase calcium alpha-sialon was reported to be 0.5<x<1.7. Recently, Yanbing Cai et al. studied the synthesis of calcium stabilized nitrogen rich alpha-sialons using CaH_2 as a starting precursor instead of Ca_3N_2 at 1800°C using hot pressing (HP) technique [28]. Single phase alpha sialon ceramics were obtained in the range of 0.50<x<1.38. Densification using conventional sintering processes, such as HP, is governed by the heat supplied via heating elements and the externally applied pressure [29]. On the other hand, for the non-conventional and modern sintering process spark plasma sintering (SPS), in addition to the uniaxial applied pressure, the pulsed nature of the current creates spark discharge and gives rise to joule heating effect between the powder particles. High rate of plasma discharge transmits and distributes the Joule heat throughout the material. This phenomenon results in a fast and uniform heat distribution within the sample, which ultimately leads to highly homogeneous samples with consistent densities. In addition, the high heating/cooling rate is beneficial as it restricts the grain growth as well as the formation of intermediate phases. Samples can be densified in very short time by SPS [29,30]. Raja et al. has reported synthesis of oxygen rich calcium alpha-sialon ceramics at relatively low temperature of 1500 °C using nano-sized precursors and SPS technique [31]. However, to our knowledge, synthesis of nitrogen rich sialons at low temperatures (below 1700°C) has not been reported in the literature.

The aim of the present study was to develop calcium stabilized nitrogen rich alpha-sialons along the Si_3N_4-1/2Ca_3N_2:3AlN line at a relatively low temperature of 1500°C using SPS and nano-/submicron-size starting powder precursors. The developed sialons were characterized for their microstructure, phase and

compositional analysis, physical and mechanical properties. Furthermore, a correlation was developed between the lattice parameters and the content of alkaline metal cation (calcium) in the alpha-sialon.

2. Experimental Procedure

Calcium stabilized nitrogen rich alpha-sialon compositions having the general formula of $Ca_xSi_{12-m}Al_mN_{16}$ (where x = m/2), along the nitrogen rich line were selected in this study. The nominal x value varied in the range of 0.2-2.2. Actually achievable compositions of alpha-sialon samples (by taking into account the oxide content on the powder surface) as listed in Table 1 were synthesized using starting powder precursors of alpha-Si_3N_4 with an average particle size of ~300 nm (Ube Industries SN-10, Japan), AlN with particle size of ~50nm (ChemPUR, Germany) and Ca_3N_2 of ~74µm (~200 mesh, Sigma Aldrich, Germany). The oxygen content in the Si_3N_4, AlN and Ca_3N_2 powders corresponds to 2.87 wt % SiO_2, 1.75 wt.% Al_2O_3 and 3.5 wt. % CaO respectively. The sample ID's used in this manuscript starts with the letter 'Ca' followed by the nominal value of 'x' selected for the synthesis of a specific composition. For example, a sample having the chemical formula of $Ca_{0.2}Si_{11.6}Al_{0.4}O_{0.52}N_{15.31}$ is referred to as Ca-0.2. Since Ca_3N_2 is very sensitive to moisture and air and easily decomposes into calcium hydroxide and ammonia, powders of silicon nitride and aluminum nitride (required to synthesize 5 gm of a specific composition) were initially mixed via ultrasonic-probe sonicator in ethanol medium. The sonicated powder mixture was then dried in oven at 95°C for a period of 24hr. to evaporate ethanol. The dried powder mixture of silicon nitride and aluminum nitride was then mixed with the required amount of calcium nitride using mortar and pestle in NEXUS II glove box (Vacuum Atmosphere Company, USA) under the presence of high purity argon gas. The powder mixture thoroughly mixed for about 15-20 minutes was then poured into a 20mm graphite die lined with BN, and the die was sealed with parafilm. The sealed die was taken out of the glove box and placed within the SPS chamber. The powder mixtures were consolidated using an SPS apparatus (FCT system, model HP D5, Germany). Regarding the calibration of the system, the schematic of the SPS is shown figure 2. During the synthesis process, the temperature was monitored and controlled using a pyrometer, fixed just above the upper punch, measuring the temperature just 3 mm away from the top surface of powder mixture inside the graphite die. Moreover, as shown in the schematic, the temperature of the die was also measured using a K-type thermocouple, placed in the center of the die, only 3 mm away from the powder sample in the die. The maximum temperature difference of ±20°C, measured between pyrometer and thermocouple, assured the in-depth calibration.

The powder mixture was consolidated into 20mm diameter pellets at a sintering temperature of 1500°C with 30 minutes holding time in vacuum environment and 50MPa of constant uniaxial pressure. The heating rate of 100°C/min was used in the synthesis, followed by constant temperature holding process, then samples were cooled down to room temperature within 5 minutes. Each sample was prepared twice and densification data was obtained from the SPS machine. The sintered samples were cleaned of graphite using SiC grinding paper and were further ground and polished using Automet 300 Buehler grinding machine to a surface finish of 1µm. Phase analysis was performed using Bruker D8 X-ray

powder diffractometer with Cu $K_{\alpha 1}$ radiation (γ = 0.15416 nm). The tube current of 40mA, accelerating voltage of 40 kV, step size of 0.02 degrees and step time of 3sec were maintained during the entire scan. In order to calculate the lattice parameters of alpha-sialon accurately, silicon powder was used as an internal standard. The amount of alpha and beta phases was calculated using the following equation [32]:

$$\frac{I_\beta}{I_\alpha + I_\beta} = \frac{1}{1 + K[(1/W_\beta) - 1]}$$

Where I_α represent observed intensities of (102) and (210) peaks belonging to alpha-phase while I_β represent observed intensities of (101) and (210) peaks belonging to beta-phase. W_β represents fraction of beta-phase while K is the proportionality constant (0.518 for β (101) and α(102) peaks while 0.544 for β(210) and α (210) peaks) [32]. The amount of minor phases was determined based on the relative intensity method. Highly polished and fractured surfaces of the synthesized samples were observed using a field emission scanning electron microscope (FESEM, Lyra 3, Tescan, Czech Republic) with an accelerating voltage of 20 kV. Phoenix focused ion beam (Helios G4 UX DualBeam, FEI) was used to prepare the lamellas for observation under transmission electron microscope (TEM). High resolution imaging and selected area diffraction patterns were acquired using FEI's Titan transmission electron microscope equipped with EDX detector. On average, five area EDX analysis were performed to calculate the cationic composition. The software UnitCellWin and XRD peak positions were used to calculate the lattice parameters. Densities of the sintered disks (after removal of graphite) was calculated using Archimedes' principle. Vickers hardness (HV_{10}) was measured on the surface of the polished samples using a universal hardness tester (Zwick-Roell, ZHU250, Germany). The indentation load of 10kg was adopted for the hardness test. Indentation method was used to calculate the fracture toughness (K_{1c}) of the sintered samples [33–36]. Evan's equation shown below was used to calculate the indentation fracture toughness. In the equation MCL stands for the maximum crack length initiated from the indentation and d is the average diagonal.

$$K_{IC} = 0.48\left(\frac{MCL}{d/2}\right)^{-1.5}\left(\frac{HV_{10}\sqrt{d/2}}{3}\right)$$

Table 1 Starting composition of calcium stabilized nitrogen rich sialons.

*Actually achievable compositions calculated by taking into account the oxide content in the starting powder precursors.

3. Results And Discussion

3.1 Sintering and Densification Behavior

Sintering behavior of the samples sintered over the range of 0.2<x<2.2 is shown in figure 3 (insert shows the complete densification curves), where the ordinate represents the displacement of upper punch of the

Sample ID	Starting Composition*	Ca_3N_2 Wt. %.	Si_3N_4 Wt. %	AlN Wt. %
Ca-0.2	$Ca_{0.2}Si_{11.6}Al_{0.4}O_{0.52}N_{15.31}$	1.74	95.37	2.89
Ca-0.4	$Ca_{0.4}Si_{11.2}Al_{0.8}O_{0.52}N_{15.32}$	3.43	90.87	5.70
Ca-0.6	$Ca_{0.6}Si_{10.8}Al_{1.2}O_{0.52}N_{15.34}$	5.08	86.48	8.44
Ca-0.8	$Ca_{0.8}Si_{10.4}Al_{1.6}O_{0.52}N_{15.35}$	6.68	82.20	11.11
Ca-1.0	$Ca_{1.0}Si_{10}Al_{2.0}O_{0.51}N_{15.36}$	8.25	78.04	13.71
Ca-1.2	$Ca_{1.2}Si_{9.6}Al_{2.4}O_{0.51}N_{15.38}$	9.78	73.98	16.25
Ca-1.4	$Ca_{1.4}Si_{9.2}Al_{2.8}O_{0.51}N_{15.39}$	11.26	70.02	18.72
Ca-1.6	$Ca_{1.6}Si_{8.8}Al_{3.2}O_{0.50}N_{15.40}$	12.71	66.15	21.13
Ca-1.8	$Ca_{1.8}Si_{8.4}Al_{3.6}O_{0.50}N_{15.42}$	14.13	62.38	23.49
Ca-2.0	$Ca_{2.0}Si_{8.1}Al_{3.9}O_{0.50}N_{15.43}$	15.51	58.70	25.79
Ca-2.2	$Ca_{2.2}Si_{7.7}Al_{4.3}O_{0.50}N_{15.44}$	16.86	55.11	28.03

die in mm, depicting the shrinkage, while the abscissa represents the sintering time. It is observed that with increase in the x value, denoting the increase in calcium (Ca) content, the shrinkage curve shifts to the left (towards shorter time). Samples having higher x value exhibit larger shrinkage (larger displacement) and subsequently were densified earlier. It is quite known that the densification of silicon nitride based materials involve the formation of a transient liquid phase, where the densification behavior depends on the viscosity; amount and wetting characteristic of the transient liquid phase. In case of the calcium stabilized alpha-sialons synthesized using oxide additives, the formation of transient liquid phase takes place as a result of reaction between silicon oxide present on the surface of silicon nitride powder and oxide additives. In our case, there are basically two sources of liquid phase. Firstly, due to the eutectic reaction between silicon oxide, aluminum oxide and calcium oxide present on the surfaces of starting nitride precursors, and secondly due to the melting of calcium nitride which takes places at about 1195°C (since calcium nitride is not used as a starting precursor in synthesis of oxygen rich calcium alpha-sialons) [37]. Furthermore, since the oxide rich eutectic liquid phase is generated as a result of surface oxide layer present on the surface of starting nitride precursors, the amount of oxide rich transient liquid is very small. On the other hand, the amount of nitrogen liquid phase formed because of melting of calcium nitride increases, with increase in x value (more calcium nitride) and consequently results in easier densification (higher shrinkage within a shorter time) of the samples having high x value. Xie.et al reported a similar trend for the nitrogen rich alpha-sialons synthesized using HP technique and micron size precursors at 1700°C and 1 hr holding time [23]. However, the reported shrinkage rate was much slower due to the slower reaction kinetics as a result of the conventional heating technique (slower heating rate) as well as due to larger size (> 1μm) of the starting powder precursors.

Table 2 shows the density values of the nitrogen rich calcium stabilized alpha-sialons for compositions having x values in the range of 0.2-2.2, synthesized at relatively low sintering temperature of 1500°C. Density of the samples sintered at 1500°C was measured to be in the range of 2.96-3.21 g/cm^3 and was seen to increase with the increase in x value (increase in Ca_3N_2 content). Xie.et al reported comparable density values (3.10-3.15 g/cm^3) with a similar trend for the nitrogen rich alpha-sialons synthesized at 1700°C using HP technique and micron size precursors [23]. Recently, Yanbing Cai et al. worked on the synthesis of calcium stabilized nitrogen rich alpha-sialons at 1800°C using HP technique [28]. The

observed density values were reported to be in the range of 3.16-3.26 g/cm^3. Furthermore, Wang et al. reported density values of 3.07g/cm^3 and 3.17g/cm^3 at 1700°C and 1750°C, respectively, for dense hot-pressed oxygen rich calcium alpha-sialon having a composition of $Ca_{0.5}Si_{10.5}Al_{1.5}O_{0.5}N_{15.5}$ [38]. Xie.et al communicated a failure to synthesize well densified yttrium stabilized nitrogen rich alpha-sialons along the Si_3N_4–YN:3AlN where insufficient amount of liquid phase was given to be the main reason [23]. This suggests that along with nano-size starting precursors and novel pulsed-based SPS technique, low melting temperature (1195°C) of calcium nitride plays a pivotal role in achieving well densified samples at 1500°C. This is in line with the literature where it has been reported that low particle size of starting powders helps sinter dense sialon materials at low temperatures [39,40].

Table 2 Density of calcium stabilized nitrogen rich alpha-sialon sintered at 1500°C.

Sample ID	Ca-0.2	Ca-0.4	Ca-0.6	Ca-0.8	Ca-1.0	Ca-1.2	Ca-1.4	Ca-1.6	Ca-1.8	Ca-2.0	Ca-2.2
X	0.2	0.4	0.6	0.8	1	1.2	1.4	1.6	1.8	2	2.2
Density (g/cm^3)	2.96	3.06	3.09	3.14	3.17	3.17	3.18	3.19	3.20	3.20	3.21

3.2 Phase Analysis

Figure 4 shows x-ray diffraction patterns of the samples synthesized at 1500°C. Single phase calcium stabilized alpha-sialon phase is observed for samples produced in the range of $0.6 \leq x \leq 1.4$ (Ca-0.6 to Ca-1.4). Dual phases (alpha and beta-sialons) are observed for samples having $x \leq 0.4$, where the amount of beta sialon phase is seen to decrease as the composition shifts towards single phase alpha-sialon. Beyond the single-phase region, alpha-sialon coexists with AlN and $CaSiAlN_3$ in the range of $1.6 \leq x \leq 2.0$. Figure 5a diplays the regenerated schematic representation of α-plane in Ca-sialon system showing phase stability regime region along the N'Rich line at 1500°C where figure 5b represents the same, where the orientation of alpha plane is the one which is most commonly presented in literature. A similar observation is reported by Xie. et al. for the calcium stabilized nitrogen rich samples synthesized at 1700°C using HP technique [23]. Furthermore, Yabaing et. al in their work on the synthesis of nitrogen rich sialons at 1800°C, have also communicated the formation of AlN and $CaSiAlN_3$ phases along with alpha-phase for higher values of x (x>1.6) and the single phase alpha-sialon phase is reported to be in the range of $0.6 \leq x \leq 1.6$ [24].

X-ray diffraction results of the samples synthesized at 1500°C are summarized in Table 3. Lattice parameters of alpha sialon are seen to increase with the increase in x value attributed to the fact that more amount of Si-N bonds are replaced by Al-N bonds as well as more amount of calcium ion is being incorporated in the structure. In calcium stabilized alpha-sialons, the charge imbalance caused as result

of substitution of Si^{+4} for Al^{+3} is brought into balance by addition of Ca^{+2} ions at the interstitial positions. It is established that alpha-sialon unit cell contains two interstices per unit cell, the upper solubility limit of calcium ion in alpha-sialon having a composition of $Ca_2Si_8Al_2N_{16}$ is ideally believed to be 2. However, due to the unavoidable presence of oxide layer on the surface of starting precursors, this composition has not been achieved so far. Recently, Yabing et al. in their work on the synthesis of nitrogen rich sialons at 1800°C, have reported a maximum achieved solubility limit of 1.82 for nominal x value of 2.2 [28]. Similarly, Xie. et al. reported the maximum solubility value of 1.7 for the calcium stabilized nitrogen rich samples synthesized at 1700°C using HP technique [23]. In our case the maximum solubility limit has been determined to be 1.83 for the nominal x value of 2.2.

Table 3 Lattice parameters, mechanical properties and phase assemblage of calcium stabilized nitrogen rich sialons synthesized at 1500°C.

Sample ID	x (Ca)		Lattice Parameters		HV_{10}	K_{1c}	Phase Assemblage
	Nom	EDXS	a (Å)	c (Å)	GPa	$MPa.m^{1/2}$	Wt.%
Si_3N_4*	0	-	7.7541	5.6217	-	-	-
Ca-0.2	0.2	0.15(2)	7.7814(6)	5.6490(4)	18.5 (3)	4.5 (4)	α (61), β (39)
Ca-0.4	0.4	0.33(3)	7.7971(6)	5.6593(4)	19.4 (7)	5.0 (4)	α (87), β (13)
Ca-0.6	0.6	0.53(3)	7.8210(7)	5.6631(5)	22.2 (2)	5.6 (3)	α (100)
Ca-0.8	0.8	0.73(3)	7.8343(7)	5.6885(5)	22.4 (5)	5.7 (3)	α (100)
Ca-1.0	1.0	0.84(4)	7.8470(7)	5.7011(5)	21.8 (4)	5.7 (3)	α (100)
Ca-1.2	1.2	1.01(5)	7.8770(7)	5.7153(4)	21.6 (2)	5.3 (2)	α (100)
Ca-1.4	1.4	1.27(3)	7.8937(7)	5.7316(5)	21.0 (3)	6.1 (3)	α (100)
Ca-1.6	1.6	1.39(4)	7.9170(7)	5.7464(5)	20.8 (4)	5.8 (2)	α (82), A (10), C (8)
Ca-1.8	1.8	1.54(5)	7.9320(6)	5.7583(5)	20.1 (7)	5.5 (3)	α (82), A (9), C (9)
Ca-2.0	2.0	1.68(4)	7.9534(7)	5.7695(5)	19.3 (6)	5.7 (2)	α (79), A (10), C (11)
Ca-2.2	2.2	1.83(5)	7.9651(6)	5.7790(4)	18.7 (5)	5.2 (3)	α (75), A (12), C(13)

*JCPDS 41-0360, (Nom) Nominal, (α) Alpha Sialon, (β) Beta-Sialon, (A) Aluminum Nitride (JCPDS 25-1133), (C) $CaSiAlN_3$ (JCPDS 39-747).

Figure 6 shows the variation of alpha sialon lattice parameters with respect to change in x value.

A similar behavior is reported by Yabning for samples synthesized at 1800°C (figure 6) [24]. The dependence of alpha-sialon lattice parameters with amount of Ca^{+2} can be well defined by the following empirical relationships:

$$a = 0.0949x + 7.7605 \text{ (Å)} \qquad (1)$$

$$c = 0.0680x + 5.6326 \text{ (Å)} \qquad (2)$$

Formation of alpha-sialon phase for sample having x value as low as 0.2 along with the observation suggesting a linear dependence of alpha-sialon unit cell parameters with x value, provide a good reason to believe that calcium stabilized nitrogen rich alpha-sialon forms continuously in the range of $0<x<1.83$. Therefore, the previously reported low solubility limit of Ca^{+2} in alpha-sialons shall not be attributed to a structural limitation but should rather be ascribed to the fact that formation of single phase alpha-sialons near the nitrogen rich corner has not been attained due to the presence of the oxide layer contamination leading to the formation of beta-sialon phase.

3.3 Microstructure Analysis

Figure 7 (a-f) depicts the secondary electron FESEM micrographs of the fractured surface of samples from the nitrogen rich sialon ceramics sintered at 1500°C. Primarily, classical intergranular fracture was observed in all samples along with some amount of grain pullout. Sample Ca-0.2 containing alpha- and beta-sialon phases exhibits elongated morphology of beta-sialon along with smaller equi-axed grains of alpha-sialon (figure 7a). Samples with x values greater than 0.4 (Ca-0.6 to Ca-2.2) composed of single phase or almost single-phase nitrogen rich calcium alpha-sialon depicted fine equi-axed grains (a morphology typical of alpha-sialon phase). However, with increase in x value (more Ca_3N_2), alpha-sialon grains with elongated morphology are also observed. Additionally, it is observed that an increase in x value yields an increase in the average grain size of alpha-sialon. The grain size distribution for the three samples (Ca-0.6, Ca-1.2 and Ca-2.0) is depicted in figure 8 where the average grain size is measured to increase from 286$\pm$ 98nm to 385$\pm$ 96 nm to 468$\pm$169 nm respectively.

Figure 9 (a-e) shows the low and high resolution TEM images of the microstructures of nitrogen rich calcium alpha-sialon samples (having the Id's as Ca-0.6, Ca-1.2 and Ca-2.0). In accordance with the observations made on fractured surfaces, sample Ca-0.6 displays an equi-axed morphology of alpha-sialon grains with an average grain size of 277 $\pm$ 34.56 nm. For samples Ca-1.2 and Ca-2.0, a dual morphology of alpha-sialon grains, i.e. elongated grains alongside the equi-axed grains, is observed. The average size of the grains is measured to be 386 $\pm$ 127nm for Ca-1.2 and 449 $\pm$ 142 nm for Ca-2.0. Alpha-sialon ceramics having an elongated morphology as a result of preferential grain growth during densification process has also been reported in the synthesis of hot-pressed oxygen rich Ca-alpha-sialons [11,38,41]. It is known that grain growth of alpha-sialon takes place via solution re-precipitation mechanism and that the presence of liquid phase is necessary to promote grain growth. Therefore, the presence of elongated morphology of alpha-sialon grains as well as the grain growth for samples having

higher x value (i.e. more Ca_3N_2) may well be attributed to the formation of a higher amount of transient liquid phase. A. Thorel et al. in their study on high temperature mechanical properties and intergranular structure of sialons (oxygen rich) have reported that intergranular phase (if present) is always amorphous and can exist as reasonably large pockets having a size in the range of 0.1 to 1μm or interfacial glassy film with a minimum reported thickness of about 60nm [42]. In our study, the high-resolution TEM images of nitrogen rich sialons (figure 9 b, d, and e) reveal that there is no sign of a glassy phase present at the grain boundaries nor at the junction of multiple grains. These results suggest that the synthesized nitrogen rich sialons would be more stable at high temperature (1400-1600^oC) compared to the oxygen rich sialons containing an amorphous grain boundary phase.

3.4 Mechanical Properties

Vickers hardness (HV_{10}) and fracture toughness (K_{1c}) of the calcium stabilized nitrogen rich alpha-sialons synthesized at 1500^oC are summarized in Table 3. The two-phase sample (Ca-0.2), having alpha- and beta-sialon phases, displays a hardness of 18.5 GPa and fracture toughness of 4.5 $MPa.m^{1/2}$. The Vickers hardness of the single-phase alpha-sialon samples (Ca-0.6 – Ca-1.4) are found to be in the range of 21.0-22.4 GPa while the fracture toughness is measured to be in the range of 5.6-6.1 $MPa.m^{1/2}$. For samples having x value greater than 1.6, a gradual decrease in the hardness is observed primarily due the formation of $CaSiAlN_3$ and AlN phases along with the major alpha-sialon phase. For a sample having 14 wt. % AlN phase and 86 wt. % $CaSiAlN_3$ phase, Y. Cai et al have reported a Vickers hardness of 14.4 GPa while Stefan R. Witek et al have reported a hardness of 12 GPa for hot pressed AlN ceramics [24,43].

Vickers hardness values in the range of 16-20 GPa and fracture toughness of 3-7 $MPa.m^{1/2}$ has been reported in the literature for the single or multi cation calcium stabilized oxygen rich alpha-sialons [11,22,38,44,45]. In our study, the relatively higher measured hardness may well be attributed to the nitrogen rich alpha-sialons. The observation is line with a study performed by František Lofaj et al. revealing that a 4% increase in N content would result in nearly the same increase in micro-hardness of RE-Si-Mg-O-N systems [46].

4. Conclusion

Calcium stabilized nitrogen rich sialons having compositions along the Si_3N_4-1/2Ca_3N_2:3AlN line were synthesized. Enhanced reaction kinetics due to nano/submicron sized precursors and high heating rates achieved with aid of non-conventional SPS process allowed the preparation of well densified nitrogen rich sialon ceramics at relatively low temperature (1500$^{\circ}C$ as compared to 1700$^{\circ}C$ or 1800$^{\circ}C$). Alpha sialon-phase for samples having Ca content (x) in the range of $0.15 < x < 1.83$ was obtained. Formation of single phase nitrogen rich alpha-sialons was achieved for samples having nominal x value in the range of $0.4 < x < 1.6$. Increase in the lattice parameters of alpha-sialon unit cell was observed to have a linear relationship with the amount of Ca^{+2} ions incorporated in the alpha structure. Increasing x value resulted in an increase in the average grain size from 286nm to 468nm. Moreover, higher amount of Ca_3N_2

starting powder was observed to facilitate the formation of elongated alpha-sialon grains. Calcium stabilized nitrogen rich sialon ceramics having very promising hardness and fracture toughness of 22.4GPa and 6.1 MPa.$m^{1/2}$ respectively were developed. Furthermore, the absence of a glassy grain boundary phase in nitrogen rich sialons make them potentially viable candidates for high temperature applications. It is believed that such engineering ceramic materials would appeal to the industry owing to their unique properties and relatively low processing temperature.

Declarations

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Compliance with Ethical Standards

Conflict of Interest: The authors declare that they have no conflict of interest.

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Figures

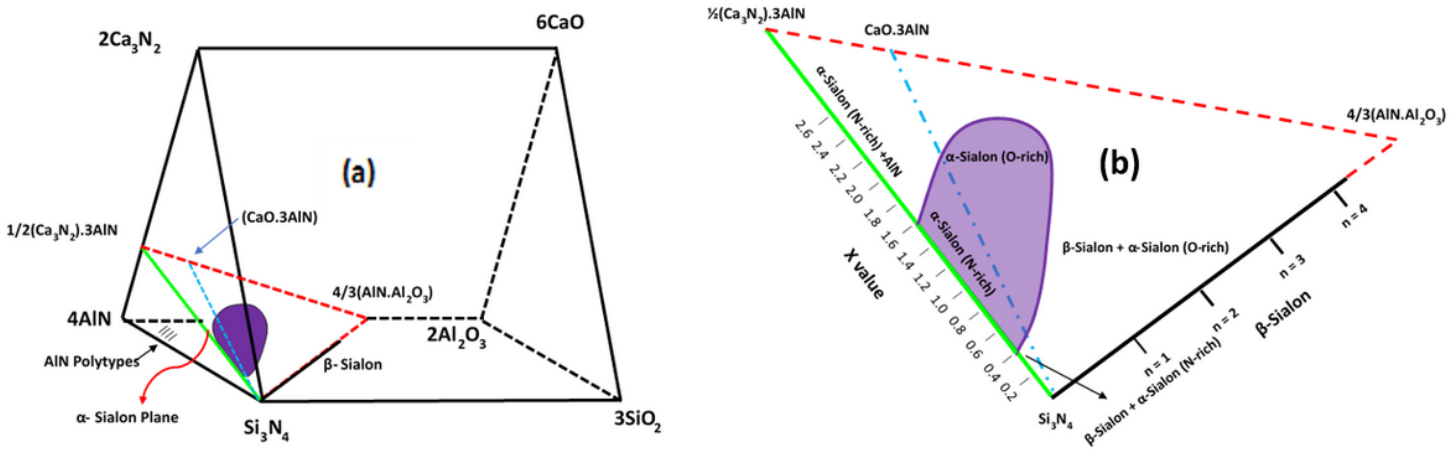


Figure 1

Regenerated schematic representation of (a) Janecke Prism of Ca-sialon system and (b) α-plane in Ca-sialon system at 1800°C [24].

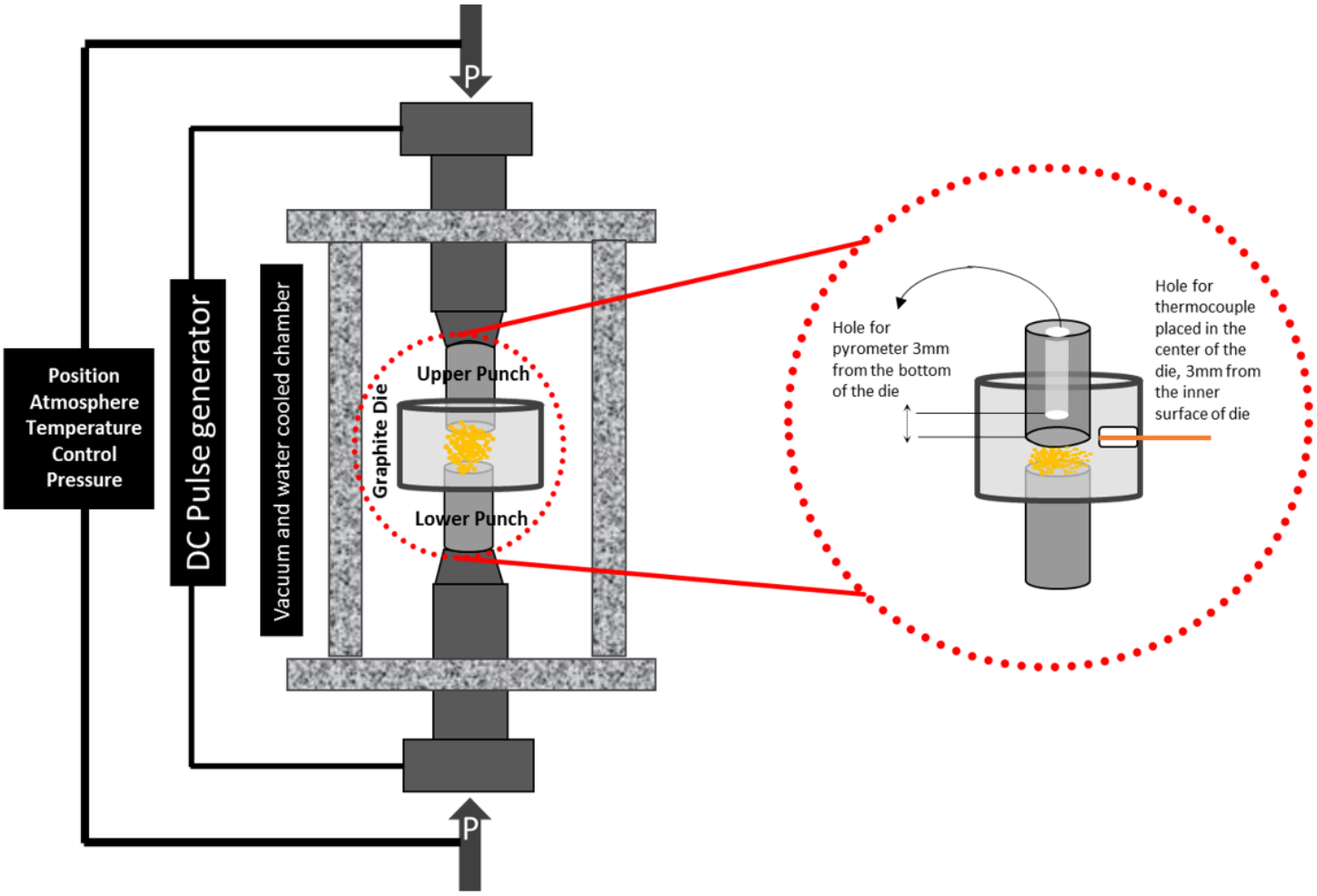


Figure 2

Schematic of spark plasma sintering equipment.

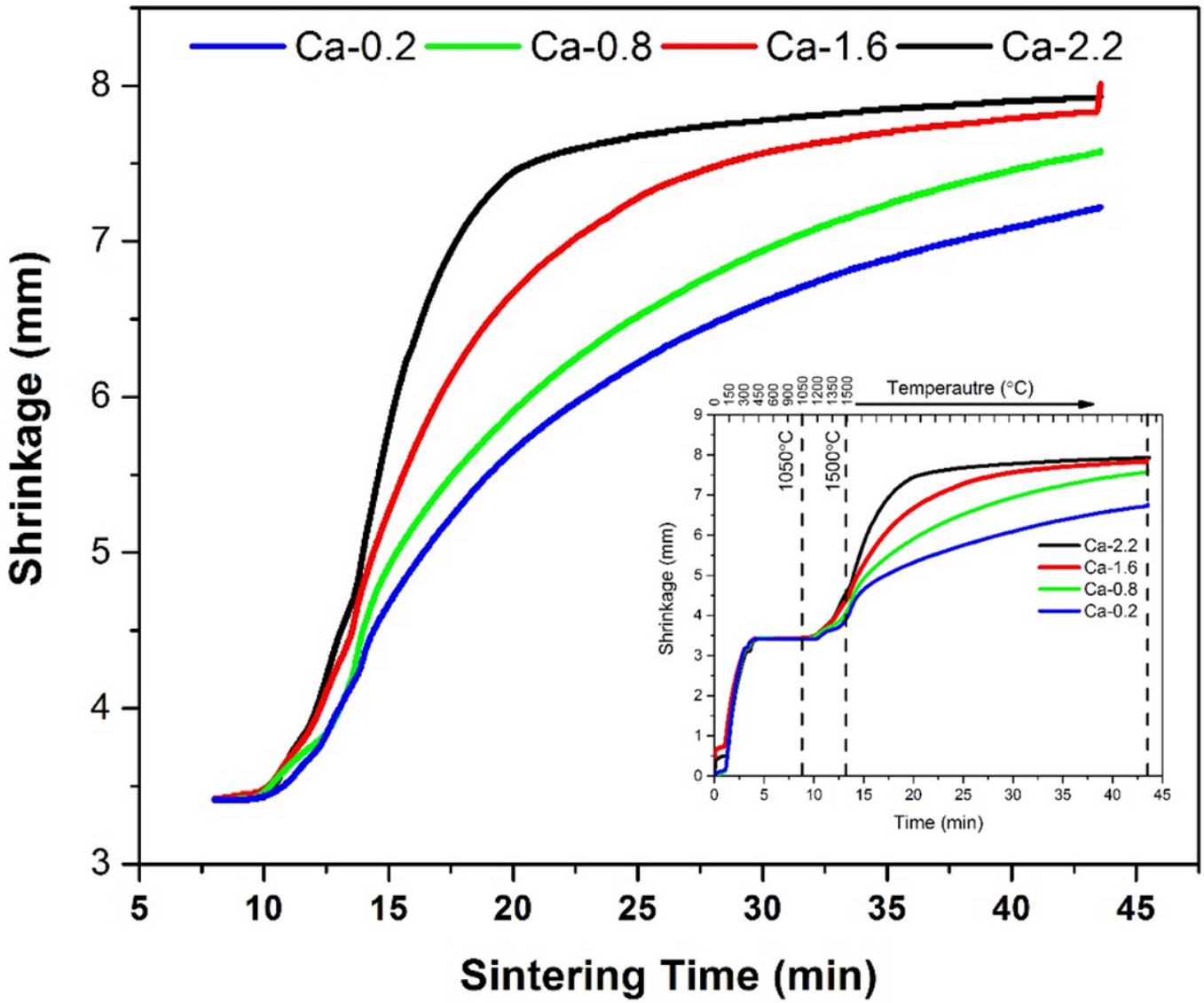


Figure 3

Densification curves of several compositions of calcium stabilized nitrogen rich sialons sintered at 1500oC using spark plasma sintering technique, where Ca content varies between 0.2-2.2. The insert shows the complete densification curves while the main plot represents the selected high temperature region (heating from 1050°C to 1500°C followed by 30 min holding time).

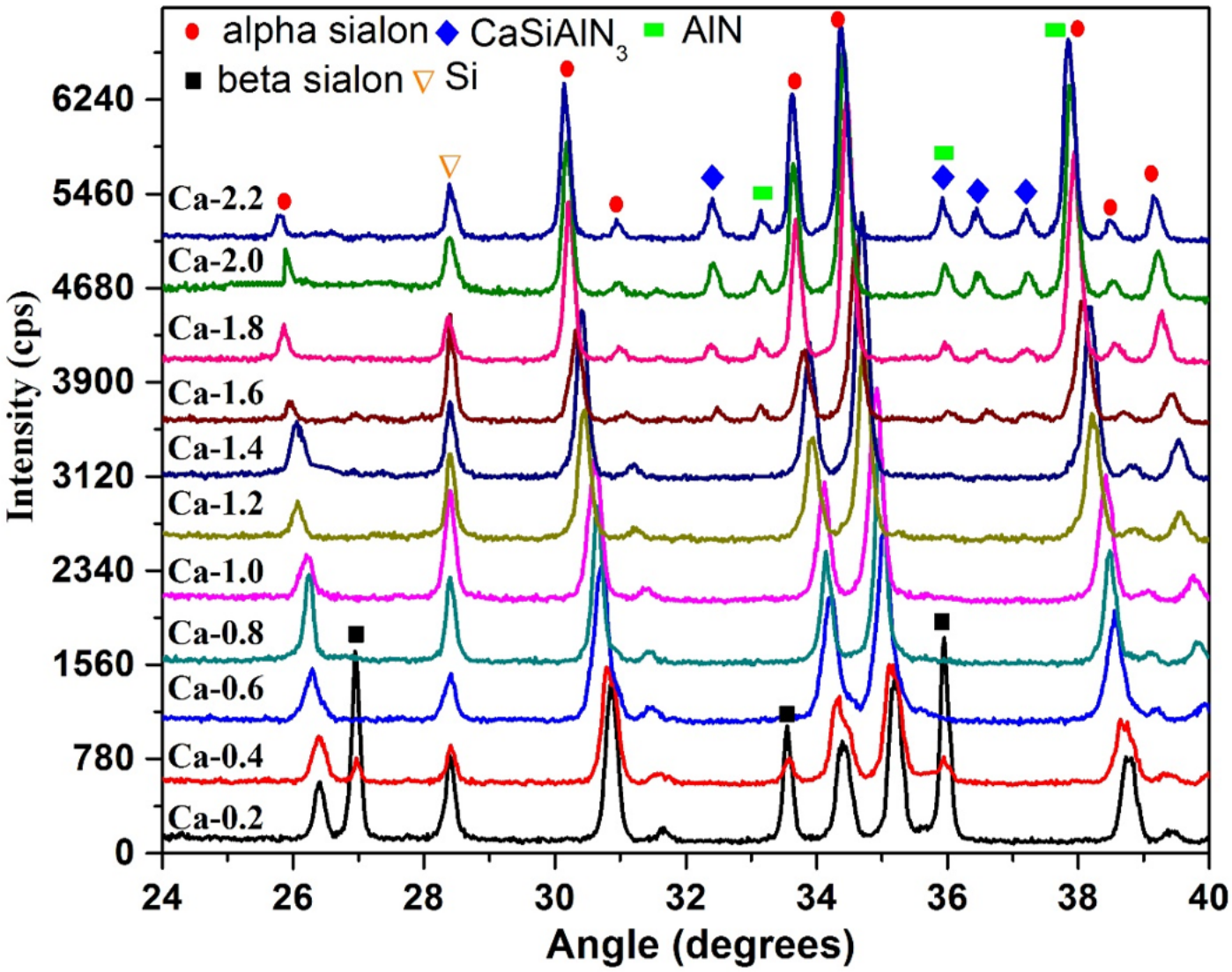
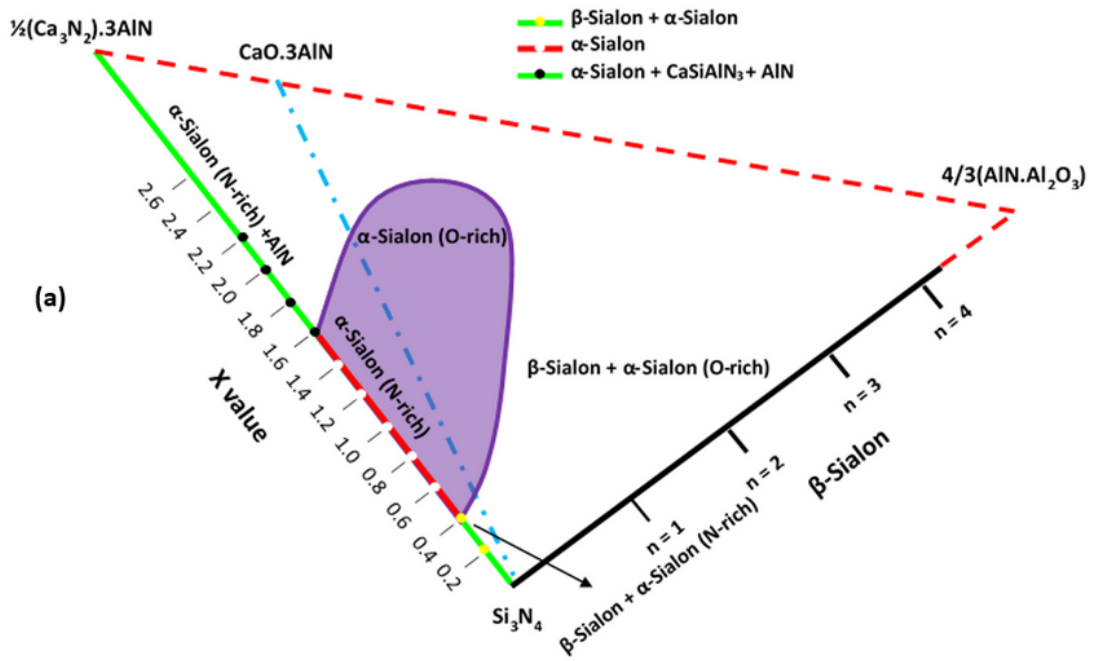


Figure 4

X-ray diffraction patterns of the calcium stabilized nitrogen rich alpha-sialon samples synthesized along the Si3N4-1/2Ca3N2:3AlN line at sintering temperature of 1500oC.



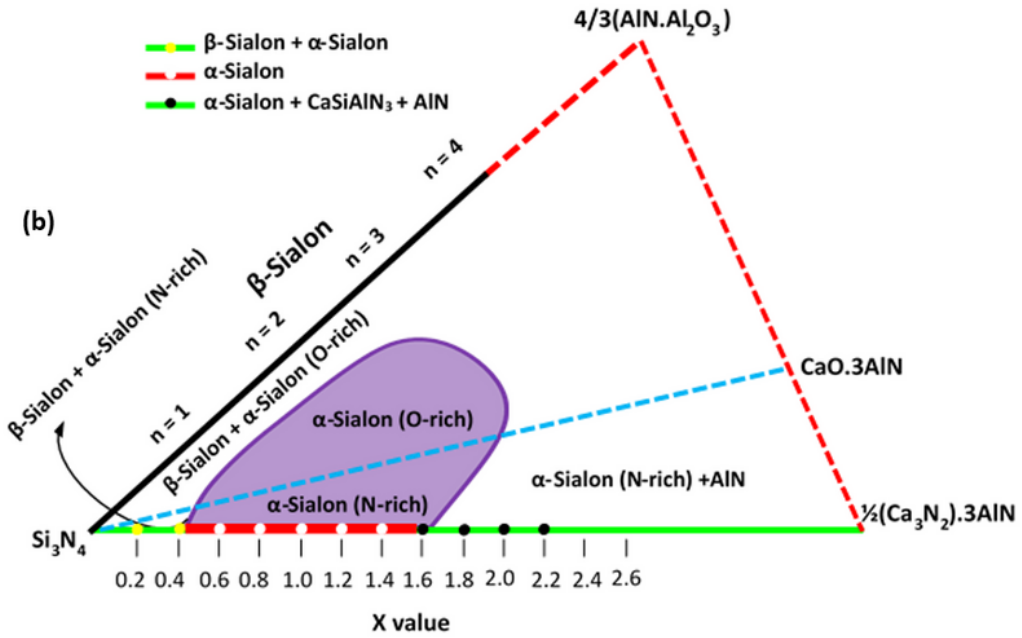


Figure 5

(a) Regenerated schematic representation of α-plane in Ca-sialon system showing phase stability regime along the N'Rich line at 1500°C and (b) another representation where the orientation of alpha plane is one which is most commonly reported in literature.

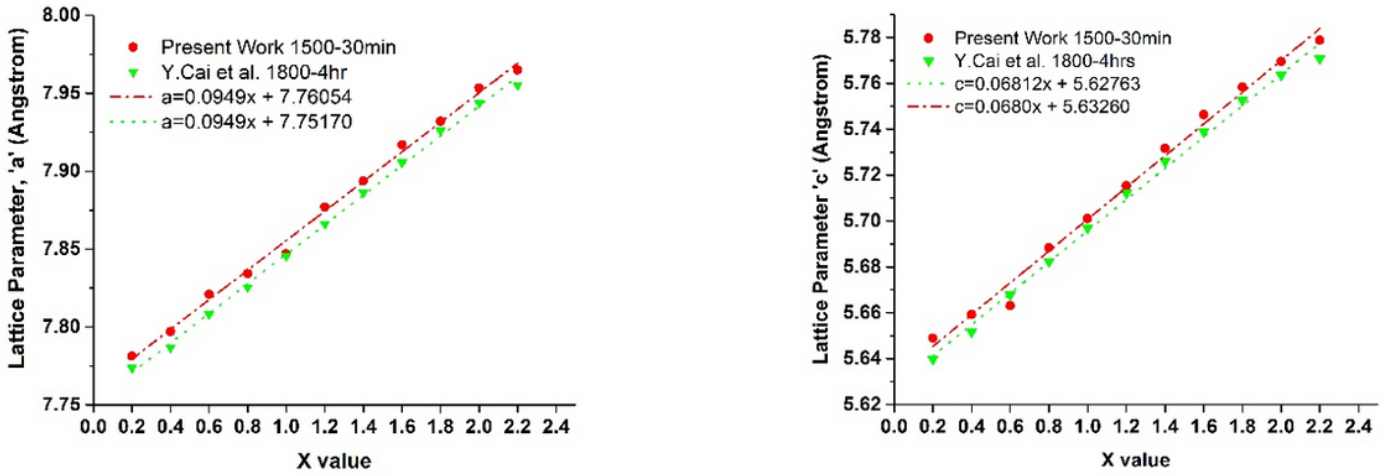


Figure 6

Variation of lattice parameters of calcium stabilized nitrogen rich alpha-sialons with x value for samples synthesized at 1500°C using SPS, and those reported by Y.Cai et al. at 1800°C using HP.

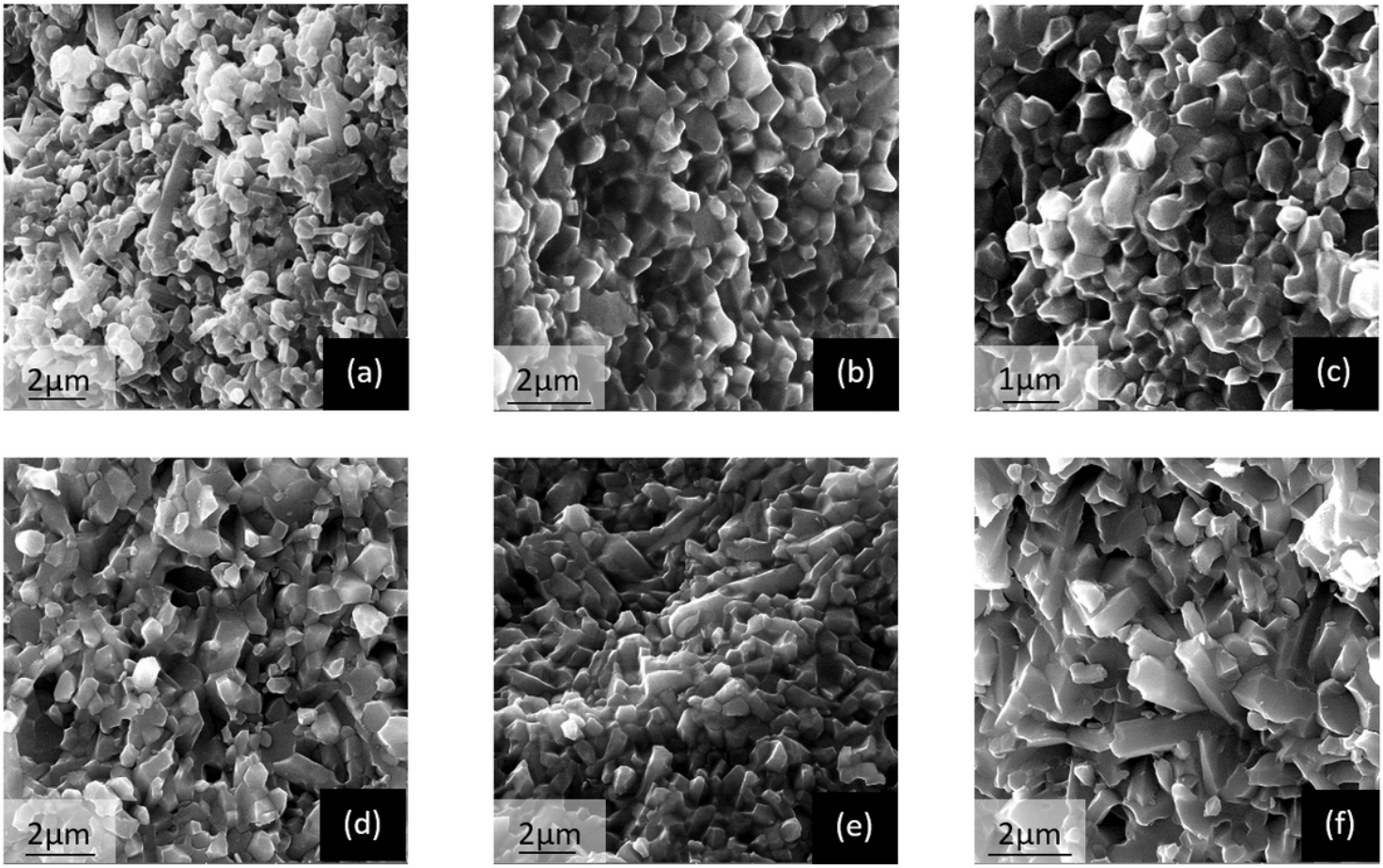


Figure 7

Secondary electron (FESEM) micrographs of the fractured surfaces of nitrogen rich samples synthesized at 1500oC using different x values: (a) Ca-0.2, (b) Ca-0.6, (c) Ca-1.2, (d) Ca-1.6, (e) Ca-1.8 and (f) Ca-2.0.

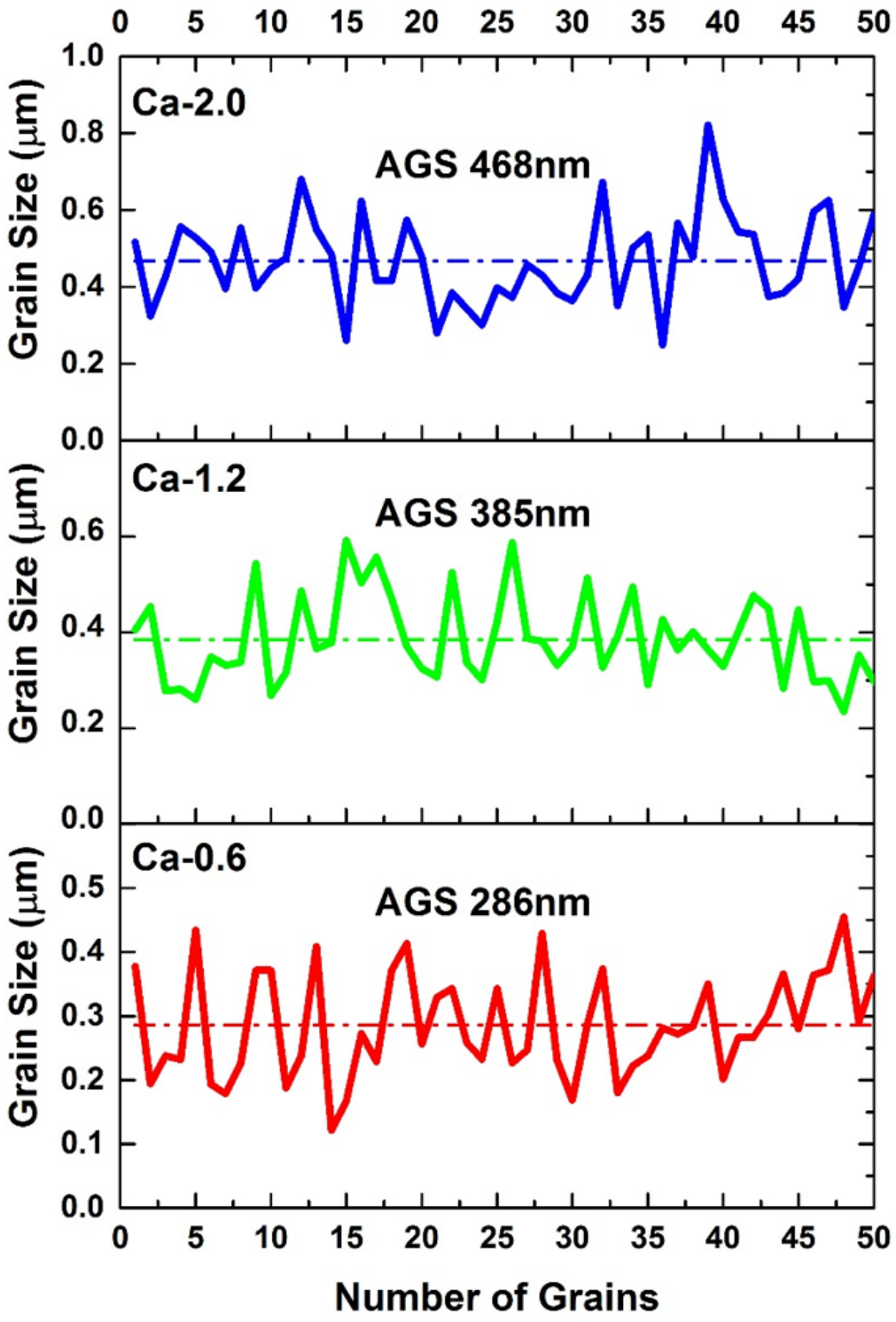


Figure 8

Grain size distribution of nitrogen rich samples synthesized at 1500oC having the sample Id's of Ca-0.6, Ca-1.2 and Ca-2.0.

Figure 9

Low and high resolution TEM micrographs of the nitrogen rich samples synthesized at 1500oC using different x values: (a & b) Ca-0.6, (c & d) Ca-1.2 and (e & f) Ca-2.0.

Supplementary Files

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