TOPICAL REVIEW

Textured silicon nitride: processing and anisotropic properties

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Abstract
Textured silicon nitride (Si₃N₄) has been intensively studied over the past 15 years because of its use for achieving its superthermal and mechanical properties. In this review we present the fundamental aspects of the processing and anisotropic properties of textured Si₃N₄, with emphasis on the anisotropic and abnormal grain growth of β-Si₃N₄, texture structure and texture analysis, processing methods and anisotropic properties. On the basis of the texturing mechanisms, the processing methods described in this article have been classified into two types: hot-working (HW) and templated grain growth (TGG). The HW method includes the hot-pressing, hot-forging and sinter-forging techniques, and the TGG method includes the cold-pressing, extrusion, tape-casting and strong magnetic field alignment techniques for β-Si₃N₄ seed crystals. Each processing technique is thoroughly discussed in terms of theoretical models and experimental data, including the texturing mechanisms and the factors affecting texture development. Also, methods of synthesizing the rodlike β-Si₃N₄ single crystals are presented. Various anisotropic properties of textured Si₃N₄ and their origins are thoroughly described and discussed, such as hardness, elastic modulus, bending strength, fracture toughness, fracture energy, creep behavior, tribological and wear behavior, erosion behavior, contact damage behavior and thermal conductivity. Models are analyzed to determine the thermal anisotropy by considering the intrinsic thermal anisotropy, degree of orientation and various microstructure factors. Textured porous Si₃N₄ with a unique microstructure composed of oriented elongated β-Si₃N₄ and anisotropic pores is also described for the first time, with emphasis on its unique mechanical and thermal-mechanical properties. Moreover, as an important related material, textured α-Sialon is also reviewed, because the presence of elongated α-Sialon grains allows the production of textured α-Sialon using the same methods as those used for textured β-Si₃N₄ and β-Sialon.

Keywords: Si₃N₄, Sialon, texture, seed, grain growth, hot-working, templated grain growth, strong magnetic field alignment, porosity, anisotropic properties

(Some figures in this article are in colour only in the electronic version)
1. Introduction

Silicon nitride (Si₃N₄) is one of the most intensively studied ceramic materials and is suitable for a variety of structural applications such as automotive engine parts, heat exchangers, pump seal parts, ball bearings, cutting tools and ceramic armor due to its low bulk density, excellent mechanical properties at elevated temperatures, high resistance to thermal shock and chemical attack, excellent creep resistance and good tribological and wear properties [1, 2].

Si₃N₄ occurs in two major crystalline phases, known as α and β, both exhibiting a hexagonal structure but with different stacking sequences: ABCD in α and ABAB in β [3]. The longer stacking sequence results in the α-phase having higher hardness than the β-phase. However, the α-phase is chemically unstable compared with the β-phase. At high temperatures (HTs) when a liquid phase is present, the α-phase always transforms into the β-phase via the solution-precipitation process, during which α-grains dissolve into the liquid phase and precipitate as β-grains, known as the α-β phase transformation [4–6]. Therefore, β-Si₃N₄ is the major form used in Si₃N₄ ceramics. In particular, its elongated growth means that β-Si₃N₄ is a well-known tough ceramic because of the unique interlocking microstructure composed of elongated grains, the so-called self-reinforced microstructure. The elongated grains can act as a reinforcing phase to increase the fracture toughness by triggering various toughening effects, such as crack bridging, crack deflection and grain pull-out [7].

Self-reinforced β-Si₃N₄ is commonly produced from an α-Si₃N₄ powder that normally contains a low fraction of β-phase. Because of the strong covalent bonding, complete densification of Si₃N₄ generally requires sintering additives such as MgO, Al₂O₃ and rare-earth (RE) oxides. During sintering, the additives react with the SiO₂ on the surface of Si₃N₄ and the Si₃N₄ itself to form a eutectic liquid phase, promoting densification, α–β phase transformation and β-grain growth. The α–β phase transformation facilitates the growth of large elongated grains, resulting in a typical bimodal microstructure composed of large elongated grains embedded in a small grained matrix [8–10], as schematically illustrated in figure 1. However, after sintering, the liquid phase invariably solidifies to an amorphous or partially crystalline secondary phase, which is located either at the grain triple junctions or the grain boundaries. The secondary phase strongly affects the thermal, physical and mechanical properties of Si₃N₄ ceramics, particularly at HTs.

The properties of Si₃N₄ ceramics can be improved by tailoring the microstructure, generally the grain morphology [9, 11–17], secondary-phase chemistry [18–38] and the grain orientation or texture [8, 39–72], as summarized in table 1. Limited fracture toughness remains a major barrier to the wide structural application of Si₃N₄ ceramics; thus, toughening has been one of the most researched topics. It has been proven that the fracture toughness can reach 8 and even 10 MPa m¹/² by increasing the diameter and volume fraction of elongated grains [11, 12, 14, 15, 17], as predicted by the crack-bridging toughening mechanism [7]. However, the large elongated grains tend to reduce bending strength by acting as large structural flaws [11, 12, 14, 15]. For example, Kawashima et al [11] reported that the development of large elongated grains of about 10µm diameter produces a high fracture toughness of 11.3 MPa m¹/² but a bending strength of only 774 MPa. To resolve this contradiction, Hirao et al [13] developed a new ‘seeding’ method involving carefully tailoring the size and amount of well-dispersed large elongated grains in a fine-grained matrix, using morphologically regulated rodlike β-Si₃N₄ single particles. They obtained self-reinforced Si₃N₄ with a fracture toughness of 8.4 MPa m¹/² and a bending strength of about 1 GPa. Using this method, Becher et al [15] showed that self-reinforced Si₃N₄ could achieve high fracture resistance (>10 MPa m¹/²) combined with a steeply rising R-curve and high fracture strength (>1 GPa).

However, the large elongated grains do not guarantee the occurrence of this toughening process [22]. For the crack-bridging toughening effect to occur, the reinforcing elongated grains must debond from the small matrix so that the crack tip is deflected along the grain face instead of splitting the elongated grains, while leaving intact elongated grains to bridge the crack [7]. The interfacial debonding process depends on the chemistry of the grain boundary phase,

![Figure 1. Scheme showing production of Si₃N₄ ceramics using α-Si₃N₄ raw powder.](image-url)
### Table 1. Microstructure design strategies and improved properties of Si₃N₄ ceramics.

| Design strategy          | Main contents | Key methods | Improved properties |
|--------------------------|---------------|-------------|---------------------|
| Grain morphology         | Diameter, aspect ratio, composition | (i) Sintering additives | Mechnical properties, thermal conductivity, oxidation resistance, etc. |
|                          |                | (ii) Sintering technique | RT–HT, thermal conductivity, oxidation resistance, etc. |
|                          |                | (iii) Seeding | Mechanical properties, thermal conductivity, oxidation resistance, etc. |
| Secondary phase chemistry| Compositional distribution and crystallization | (i) Templated grain growth | Mechanical properties, RT–HT, thermal conductivity, oxidation resistance, etc. |
|                          |                | (ii) Hot-working | Mechanical properties, RT–HT, thermal conductivity, oxidation resistance, etc. |

| Representative examples² | |
|--------------------------|-----------------|
| 1994, Hirao et al. [13] | Seeded Si₃N₄:  σᵧ = 40 MPa m¹/₂, K_c = 10 MPa m¹/₂. |
| 2001, Hayashi et al. [16] | Yb₂O₃-MgSiN₂-doped Si₃N₄: σᵧ = 140 MPa m¹/₂, K_c = 14 MPa m¹/₂. |
| 2001, Guo et al. [21] | Lu₂O₃-doped Si₃N₄: σᵧ = 660 MPa at 1500 °C. |
| 2006, Zeng et al. [27] | Lu₂Si₃O₇N₂ [24–26, 32–37]; and Lu₂Si₃O₇N₂ [31, 34, 37] grain boundary phases. |

³ RT = room temperature, HT = high temperature.
⁴ _∥_ and _⊥_ mean the directions parallel and perpendicular to the axis of elongated β-Si₃N₄ grains for textured Si₃N₄.
composition (Lu₂O₃–SiO₂), Zeng et al [72] reported the formation of textured Si₃N₄ exhibiting a bending strength of \( \approx 740 \) MPa at both room temperature (RT) and 1500 °C along the direction perpendicular to the grain alignment, that is, no strength degradation occurs even at 1500 °C, indicating better HT strength retention than that of untextured materials. Moreover, textured Si₃N₄ exhibits unique anisotropic thermal conductivity [39, 45, 47, 52, 53, 58, 59, 62, 65, 72], tribological and wear properties [54, 55, 63, 66] and erosion behavior [67, 69]. For example, Watari et al [52] showed that thermal conductivities of 155 and 52 Wm⁻¹ K⁻¹ could be achieved in highly textured Si₃N₄ ceramic along the directions parallel and perpendicular to the grain alignment, respectively.

To date, a large number of publications are available reporting research on each microstructure design strategy. There have been several review articles reporting the progress in research on grain morphology [73–75] and secondary-phase chemistry [76]. However, there is no full review article available concerning the research progress in the field of textured Si₃N₄ ceramics. Because texturing offers a unique opportunity to produce super-Si₃N₄ ceramics, a considerable amount of research has been carried out, leading to rapid progress in the processing and anisotropic properties of textured Si₃N₄ ceramics, which are expected to be used for a broad range of applications in extremely severe environments. Therefore, the objective of this article is to review research on textured Si₃N₄ ceramics, emphasizing the following areas: (i) grain growth behavior of \( \beta \)-Si₃N₄, (ii) typical texture structure and analysis, (iii) processing methods for textured Si₃N₄, (iv) anisotropic properties and (v) processing and properties of textured porous Si₃N₄. As an important related material, textured \( \alpha \)-Sialon is also reviewed in this article.

2. Definition of textured Si₃N₄

There are two major forms of textured ceramics related to Si₃N₄ in the literature. One form comprises Si₃N₄ composites reinforced by secondary-phase particles or whiskers, such as Si₃N₄–BN [77–79], Si₃C₆/Si₃N₄ [80] and Si₃C₇/Si₃N₄ [81]. In these composites, the texture results from the orientation of either elongated \( \beta \)-Si₃N₄ grains or secondary-phase particles (or whiskers) or both of them. The other form is self-reinforced Si₃N₄. Here, the texture results only from the preferential orientation of elongated β-Si₃N₄ grains. In this article, we focus only on the second form, the so-called textured Si₃N₄. Therefore, the term ‘textured silicon nitride (Si₃N₄)’ is defined as a self-reinforced silicon nitride with the preferential orientation of elongated \( \beta \)-Si₃N₄ grains. In the literature, another name, ‘anisotropic silicon nitride’, has also frequently been used for this material. In addition, it is worth emphasizing that textured \( \beta \)-Si₃N₄ in practice refers to textured \( \beta \)-Si₃N₄.

3. Grain growth of \( \beta \)-Si₃N₄

To help with understanding the texturing behavior in Si₃N₄ ceramics, we wish to briefly outline \( \beta \)-Si₃N₄ grain growth behavior, including anisotropic grain growth and abnormal grain growth, both of which play a critical role in the microstructural evolution and properties of Si₃N₄ ceramics.

3.1. Anisotropic grain growth

\( \beta \)-Si₃N₄ grains are known to grow preferentially into a typically hexagonal rodlike shape during liquid-phase sintering of either the \( \alpha \) or \( \beta \) powder. This is attributed to the much faster growth rate along the [001] (or c-axis) direction than along the [210] direction, namely so-called anisotropic or elongated grain growth. The elongated grain growth allows Si₃N₄ ceramics to exhibit a unique self-reinforced microstructure and also a textured microstructure. Figure 2 shows (a) a typical texture microstructure containing aligned elongated \( \beta \)-Si₃N₄ grains along the c-axis direction [82] and (b) a schematic of a hexagonal rodlike \( \beta \)-Si₃N₄ grain growing from an initial nucleus or preexisting seed crystal. The white arrow indicates a typical large elongated grain growing epitaxially from an initial seed particle (dark gray color) along the c-axis direction during sintering, the so-called core-rim or shell structure.

![Figure 2](image)

Figure 2. Illustration showing typical anisotropic grain growth of \( \beta \)-Si₃N₄: (a) textured microstructure with aligned elongated \( \beta \)-Si₃N₄ grains along the c-axis and (b) schematic of an elongated hexagonal \( \beta \)-Si₃N₄ grain showing anisotropic grain growth in the [001] and [210] directions. The white arrow indicates a typical large elongated grain growing epitaxially from an initial seed particle (dark gray color) along the c-axis direction during sintering, the so-called core-rim or shell structure.

3.1.1. Rough–smooth plane model. On the basis of the periodic bond chain (PBC) analysis of \( \beta \)-Si₃N₄, Krämer et al [85] proposed a model that relates anisotropic grain growth to the crystal structure of \( \beta \)-Si₃N₄, in which the basal (001) plane is atomically rough but the prism (100) plane is atomically smooth. This prediction was also supported by a published transmission electron microscopy (TEM) study [86]. The rough basal plane allows a lower activation energy for nucleation and thus continuous growth...
in comparison with the smooth prism plane. Thus, this unique interface characteristic leads to the faster growth kinetics in the [001] direction than in the [210] direction, and thereby the observed anisotropic grain shape of β-Si3N4. On the basis of this model, Hwang et al [86] provided a reasonable explanation of the anisotropic grain growth of β-Sialon by developing new growth kinetics equations as an alternative to the classic Ostwald ripening model:

\[ d^n - d^n_0 = k t, \]

where \( d_0 \) is the initial mean grain size, \( d \) is the mean grain size at time \( t \) and \( k \) is the rate constant. The Lifshitz–Slyozow–Wagner (LSW) theory predicts growth exponents of \( n = 2 \) and 3 for interfacial reaction control and diffusion control, respectively. In their work, they found that \( n = 3 \) for both length [001] and width [210] directions, indicating diffusion control. However, Lai and Tien [87] observed that \( n = 3 \) and 5 for the length and width directions of β-Si3N4 in sintered Si3N4 ceramics, respectively. The higher calculated activation energy for the width direction than for the length direction appears to support the rough–smooth plane model.

3.1.2. Anisotropic Ostwald ripening model. The classic Ostwald ripening model is not applicable for describing the anisotropic grain growth of β-Si3N4 because of the use of only one parameter, grain size, \( d(t) \); thus, Kitayama et al [88] developed a thermodynamic model for anisotropic Ostwald ripening using two parameters, \( w(t) \) and \( l(t) \), for the growth of β-Si3N4 in the width and length directions, respectively. In this model, the chemical potential difference between facets and the liquid phase is the driving force for mass transport.

This mode reveals that the reduction of the aspect ratio after the \( \alpha \rightarrow \beta \) phase transformation observed experimentally is a consequence of anisotropic Ostwald ripening, and the length is a complicated function of time and the relationship between the diffusion and the interfacial reaction constants. This model predicts a growth exponent of \( n = 3 \) for a interfacial reaction controlled by kinetics, and higher values when the diffusion constants approach the interfacial reaction constants, both of which are consistent with the observed experimental results in the literature, whereas the exponent of \( n = 2 \) has never been observed for β-Si3N4, as predicted by the LSW theory. The same authors further extended this model to predict the grain growth during the \( \alpha \rightarrow \beta \) phase transformation [89] and the tip shape evolution of a β-Si3N4 crystal in the liquid phase [90]. The results demonstrated that the ratio of the interfacial reaction constants of the (100) and (001) planes and the \( \alpha \rightarrow \beta \) ratio are the key factors determining the anisotropic grain growth of β-Si3N4.

The anisotropic Ostwald ripening model provides a successful explanation of why the effect of lanthanide additives on the anisotropic grain growth of β-Si3N4 is due to the change in the interfacial reaction constants between the (100) facets of β-Si3N4 crystals and the Ln–Si–O–N liquid phase [83].

3.1.3. Acid-base model. On the basis of the interfacial segregation mechanism of cations, Wang et al [91] proposed an acid-based model to explain the effects of the liquid-phase composition on the anisotropic grain growth of β-Si3N4. In this model, silicon and aluminum are a strong acid and weak acid, respectively, and the basicity of yttrium and RE bases increases as the ionic size increases. Again, this model treats the Si3N4 crystal as a strong acid, the (Si, Al, M)(O, N) liquid (M represents the network modifiers) as a weak acid and the modifier cations as strong bases that tend to segregate to the prismatic faces of the crystal. Thus, the anisotropic grain growth rate depends on the difference in the acidity between the crystal and the liquid. This model suggests that (i) the strong segregation inhibits the growth of prismatic planes relative to the basal plane and (ii) the anisotropic grain growth is enhanced as the ionic radius of the RE element increases because of the intensified segregation.

3.1.4. Differential binding energy model. The β-Si3N4 grain morphology is strongly affected by RE oxides, the most commonly used additives for sintering Si3N4 ceramics. However, the anisotropic grain growth behavior is difficult to rationalize in terms of the ionic radius of the RE cations or the cationic field strength. Painter et al [92] developed a differential-binding-energy (DBE) model to explain the role of RE oxides in the anisotropic grain growth of β-Si3N4. The DBE characterizes the energy contribution to the RE segregation due to the competition between the RE and Si for anion bonding. By selecting sixfold and fourfold coordinated O and N as reference hosts, respectively, the DBE is defined as:

\[ \delta_{\text{Si},\text{RE}} = (\Delta E_{\text{RE}O_6} - \Delta E_{\text{RE}N_4}) - (\Delta E_{\text{Si}O_6} - \Delta E_{\text{SiN}_4}), \]

where \( \Delta E_{\text{IJ}} \) is the binding energy of cation I (Si or RE), in an anion J (O or N) environment, which models the oxygen- or nitrogen-rich host sites in the ceramics. For an RE in an octahedral O environment, the binding energy is

\[ \Delta E_{\text{RE}O_6} = E_{\text{RE}O_6} - E_{\text{O}_6} - E_{\text{RE}}, \]

where \( E_K \) is the calculated total energy for the relaxed system K.

This model describes a physical mechanism in which the anisotropic grain growth originates from the site competition between REs and Si for bonding at β-Si3N4 interfaces and within the O-rich glass. This model gives a successful explanation of the observed variation of the anisotropic grain growth depending on the RE oxide [93], as illustrated in figure 3. In contrast with Lu, which has the highest DBE value, La has the lowest DBE value, leading to the strongest preferential segregation and highest binding strength to the prismatic grain surface, thereby the strongest anisotropic grain growth represented by the highest aspect ratio.

3.2. Abnormal grain growth
To toughen Si3N4 ceramics, a major processing strategy is to promote the development of large elongated β-Si3N4 grains or abnormal grain growth, thereby resulting in the desired bimodal microstructure. However, the large elongated
grains tend to reduce the strength of the materials, and the control of abnormal grain growth is crucial for the realization of high-performance Si₃N₄ ceramics. Dressler et al. [94] demonstrated that the abnormal grain growth of β-Si₃N₄ is governed by the morphology, the grain-size distribution and the amount of β-Si₃N₄ nuclei present in the α-Si₃N₄ powder. The same authors proposed a model to explain the effect of β-Si₃N₄ nuclei size on abnormal grain growth, as shown in figure 4. This model predicts the dissolution of small β-grains if they are located within the diffusion gradient of a large β-grain. The authors suggest that the use of faceted and elongated β-Si₃N₄ nuclei crystals with a narrow size distribution is beneficial for obtaining the desired bimodal microstructure that gives Si₃N₄ ceramics both high strength and high fracture toughness.

Moreover, Emoto and Mitomo [95] demonstrated that the abnormal grain growth of β-Si₃N₄ is also governed by the morphology, the grain-size distribution and the amount of β-Si₃N₄ nuclei in the β-Si₃N₄ powders, where the α→β phase transformation is not involved. The driving force for abnormal grain growth, ΔC, is defined as

\[ \Delta C = \Delta C_i + \Delta C_p + \Delta C_c, \] (4)

where ΔCᵢ is the difference in solubility based on the particle size (the solubility of a smaller spherical grain is higher than that of a larger grain of the same phase, ΔCₚ is the crystal phase (the solubility of a spherical α-grain is higher than that of a β-grain of the same size and shape), and ΔCᵢ is the interfacial energy (the solubility of a spherical grain is higher than that of a faceted grain of the same phase and volume). If a critical driving force exists, ΔCᵢ + ΔCₚ + ΔCᵢ, which causes abnormal grain growth; when ΔC > ΔCᵢ, abnormal grain growth occurs, but when ΔC < ΔCᵢ normal grain growth occurs.

In the literature, there are several major processing methods available reported for controlling abnormal grain growth in Si₃N₄ ceramics, including

(i) seeding with single β-Si₃N₄ crystal particles [13–15],
(ii) sintering/annealing at high temperature for a long time [11, 12, 16, 17],
(iii) additive composition including composition and amount [16, 96, 97],
(iv) preheating composition before final sintering [98, 99].

4. Texture structure and analysis

4.1. Texture structure

Textured Si₃N₄ occurs in two major microstructural forms: α-, β-axis orientation and c-axis orientation. The α-, β-axis oriented texture forms as a result of the orientation of the prism planes of elongated grains. This orientation is also called the planar orientation. However, the c-axis oriented texture forms as a result of the orientation of elongated grains along the long-axis (or c-axis) direction. This orientation is also called the axial orientation.

Figures 5(a) and (b) show typical features of the three-dimensional (3D) α-, β-axis oriented texture, in which elongated β-Si₃N₄ grains have planar orientation, but are randomly oriented along the c-axis (or long-axis) direction. This texture can be easily identified by standard powder x-ray diffraction (XRD) analysis, as shown in figure 5(c). On the planes parallel to the c-axis, the diffraction peaks of the (h00) planes parallel to the c-axis of the grains are substantially stronger, typically similar to those of (200) and (210) planes,
and the (002) plane perpendicular to the $c$-axis disappears. However, on the planes parallel to the $a, b$-axis, the diffraction peaks of the $(hk0)$ planes become relatively weak, whereas the diffraction peak of the (101) plane intersecting the $c$-axis is the strongest, and the diffraction peak of the (002) plane is enhanced. The degree of orientation in such a texture can be intuitively evaluated from the relative peak intensity of the (101) plane in the XRD pattern of the top plane ($\parallel c$-axis). On the whole, the degree of orientation increases as the (101) peak intensity decreases.

Figures 6(a) and (b) show typical features of the 3D $c$-axis oriented texture, in which elongated $\beta$-Si$_3$N$_4$ grains are unidirectionally oriented along the $c$-axis direction. This texture can also be easily identified by XRD analysis, as shown in figure 6(c). On the planes parallel to the $c$-axis, the diffraction peaks of the $(hk0)$ planes exhibit stronger relative intensities, whereas those of the (101) and (002) planes are very weak. However, on the planes perpendicular to the $c$-axis (also parallel to the $a, b$-axis), the diffraction peaks of the $(hk0)$ planes are weak, whereas the diffraction peaks of the (101) and (002) planes are the strongest. The relative intensity of the diffraction peaks between the (101) and (002) planes depends on the degree of orientation along the $c$-axis direction. The higher orientation allows the (002) peak to exhibit stronger intensity than the (101) peak. If the (101) peak is not detected, a perfect $c$-axis aligned texture is considered to have developed.

4.2. Texture analysis

The texture analysis methods reported for Si$_3$N$_4$ ceramics in the literature include the following.

4.2.1. X-ray diffraction method. XRD is the most widely used method for identifying and analyzing the texture development in materials. Using the XRD data, the following methods can be used to analyze the texture in Si$_3$N$_4$, including.
a. Relative peak intensity

Relative peak intensity is a simple and direct qualitative method for identifying the orientation of Si3N4 grains in green compact and sintered samples, particularly in green compacts with very weak orientation because of the very limited amount of β-Si3N4, the most common case for α-Si3N4 raw powder. The relative peak intensity ratios mostly used for β-Si3N4 include \( I_{(210)}/I_{(101)} \), \( I_{(200)}/I_{(101)} \), \( I_{(220)}/I_{(101)} \) and \( I_{(400)}/I_{(101)} \). For example, the \( a, b \)-axis orientation (figure 5) can be identified by the higher \( I_{(210)}/I_{(101)} \) value on the plane parallel to the c-axis than that on the plane parallel to the \( a, b \)-axis, as shown in figure 5(c). Nevertheless, when Y2O3 is used as a sintering additive, the texture identification of α-Si3N4 green compact, produced by strong magnetic field alignment (SMFA) using slip casting (discussed later), encounters a green compact, produced by strong magnetic field alignment (SMFA) using slip casting (discussed later), encounters a problem that the \( (400) \) and \( (411) \) peaks of Y2O3 overlap with \( (002) \) using slip casting (discussed later), encounters a green compact, produced by strong magnetic field alignment.

\[ I_{(210)} = \frac{I_{(210)}}{I_{(101)}} = \frac{I_{(002)}}{I_{(101)}} \times 0.05 \]  

(5)

\[ I_{(101)} = \frac{I_{(101)}}{I_{(222)}} = \frac{I_{(333)}}{I_{(400)}} \times 0.23 \]  

(6)

In equations (5) and (6), the factors 0.05 and 0.23 result from the calculated values of \( I_{(411)}/I_{(222)} \) and \( I_{(400)}/I_{(222)} \), from the measured XRD peak data of the cast sample in the absence of a magnetic field, respectively. This is in agreement with the data on the standard PDF card (No. 41-1105), showing that the values of \( I_{(400)}/I_{(222)} \) and \( I_{(411)}/I_{(222)} \) are 0.24 and 0.05, respectively.

b. Orientation indices

On the basis of XRD peak intensity data, the texture in β-Si3N4 ceramics can be evaluated by the orientation index of the \((hkl)\) plane for β-Si3N4, \( N_{h,k,l} \), which is defined as [101]

\[ N_{h,k,l} = \frac{F_{h,k,l}}{F_{h,k,l}^0} \]  

(7)

\[ F_{h,k,l} \text{ and } F_{h,k,l}^0 = \frac{I_{h,k,l}}{\sum_{i=1}^{n} I_{h,k,l}} \]  

(8)

where \( I_{h,k,l} \) is the diffraction peak intensity of the \((hkl)\) plane, and \( n \) is the number of diffraction peaks. \( F_{h,k,l}^0 \) is calculated from data on the standard PDF card (No. 33-1160) of β-Si3N4.

c. Relative facial angle

A relative facial angle measured from the \( c \)-plane, \( \theta_F \), has been proposed to evaluate the degree of crystalline orientation from the intensity of XRD peaks. \( \theta_F \) is defined as [102]

\[ \theta_F = \frac{\sum (I_{h,k,l} \times \theta_{h,k,l})}{\sum I_{h,k,l}} \]  

(9)

where \( \theta_{h,k,l} \) is the facial angle between the \((hkl)\) and \((001)\) planes, \( I_{h,k,l} \) is the peak intensity of the \((hkl)\) plane obtained from the XRD pattern. The facial angle \( \theta_F \) is reduced to 0° when all crystals are oriented to the c-plane and to 90° when they are oriented to the \( a, b \)-plane. Li et al [103] reported that the static magnetic field allows the β-Si3N4 grains in sintered Si3N4 to exhibit a \( \theta_F \) of 80°, suggesting the \( a, b \)-axis orientation of β-Si3N4 grains.

d. Lotgering orientation factor

The Lotgering orientation factor has been widely used to evaluate the degree of texture in ceramics because of its simplicity, speed and inexpensiveness. The Lotgering orientation factor, \( f_L \), can be expressed as [104]

\[ f_L = \frac{P - P_0}{1 - P_0} \]  

(10)

For \( a, b \)-axis orientation,

\[ P = \frac{\sum I_{(h,k,0)}}{\sum I_{(h,k,l)}} \]  

(11)

and for \( c \)-axis orientation,

\[ P = \frac{\sum I_{(0,0,l)}}{\sum I_{(h,k,l)}} \]  

(12)

where \( \sum I_{(h,k,0)} \) and \( \sum I_{(0,0,l)} \) are the sums of peak intensities of the \((h00)\) and \((00l)\) planes parallel and perpendicular to the \( c \)-axis of β-Si3N4 crystal, respectively, and \( \sum I_{(h,k,l)} \) is the sum of peak intensities of all the \((hkl)\) planes in the range of 2\( \theta \). The values of \( P \) are obtained from the sample, and the values of \( P_0 \) are obtained from the standard PDF card (No. 33-1160) of β-Si3N4. If \( f_L = 0 \), no grain orientation occurs; if \( f_L = 1 \), perfect orientation is considered to have developed. A larger absolute value of \( f_L \) implies a higher orientation.

However, the Lotgering orientation factor does not provide quantitative information on the degree of misorientation of grains or the textured volume fraction, thus, it allows only a semiquantitative analysis. Lotgering [104] pointed out that the accuracy of this method decreases with decreasing orientation. In addition, it was reported that the value of \( f_L \) is dependent on the number of reflections used in the calculation [105]. To determine this effect on the Lotgering orientation factor of textured Si3N4, we conducted calculations for various ranges of 2\( \theta \), as shown in table 2. Our results suggest that the 2\( \theta \) range has no significant effect on the \( f_L \) value, even for lower orientation, and the effect is less than that observed by Jones et al in bismuth titanate ceramics [105]. This calculation may be helpful for shortening the experimental time when using this method to determine the degree of texture in Si3N4 ceramics.

e. Pole figure

A pole figure is a common and effective quantitative method for evaluating the texture development in materials [106]. A pole figure provides information on the distribution of...
Figure 7. Typical (a) (200) and (b) (002) pole figures measured from textured \( \beta \)-Si\(_3\)N\(_4\) prepared by PSC (reproduced with permission from [107] ©2000 Blackwell Publishing Ltd) and TCA techniques (reproduced with permission from [108] ©2003 Blackwell Publishing Ltd), corresponding to highly \( a \), \( b \)-axis and \( c \)-axis oriented \( \beta \)-Si\(_3\)N\(_4\), respectively. The number gives the multiple of random distribution (mrd), indicating the degree of preferential orientation. Both pole figures are perpendicular to the paper, i.e., the HF direction is located at the center of the (200) pole figure, and the sheet-stacking direction is located at the center of the (002) pole figure.

Table 2. Effect of \( 2\theta \) range on the calculated Lotgering orientation factor in \( a \), \( b \)-axis and \( c \)-axis oriented Si\(_3\)N\(_4\).

| \( 2\theta (^\circ) \) | \( a \), \( b \)-axis orientation | \( c \)-axis orientation |
|---------------------|-------------------------------|-------------------------|
| HD                  | MD                            | LD                      |
| 10–70               | 0.97                          | 0.18                    |
| 20–55               | 0.97                          | 0.18                    |
| 30–70               | 0.50                          | 0.18                    |

HD = high degree, MD = medium degree, LD = low degree of orientation of \( \beta \)-Si\(_3\)N\(_4\) grains.

4.2.2. Scanning electron microscopy (SEM) image analysis. On the basis of the image analysis of SEM micrographs, Imamura et al [44] developed a formula for determining the degree of orientation of elongated \( \beta \)-Si\(_3\)N\(_4\) grains, defined as

\[
f_I = 1 - \frac{\psi}{180^\circ} \quad (0 \leq \psi \leq 180^\circ) \quad (13)
\]

where \( \psi \) is defined as the half bandwidth of the peak in the curve of the area fraction of elongated grains plotted against angle \( \theta \) \((-90^\circ \leq \theta \leq 90^\circ)\) between the long axis of the grains and the tape-casting direction. Moreover, the grain orientation in Si\(_3\)N\(_4\) ceramics can be quantitatively evaluated using an orientation factor [114], \( f_P \), which has been used for describing the planar state of fiber orientation in a discontinuous fiber-reinforced composite, defined as [114]

\[
f_P = \frac{2}{N} \sum_{i=1}^{N} \cos^2 \theta_i - 1 \quad (14)
\]

where \( N \) is the number of grains, and \( f_P = 0 \) for the random orientation and \( f_P = 1 \) for the perfect orientation.

4.2.3. Electron backscattered diffraction (EBSD) method. The EBSD method has been utilized to analyze the texture in \( \beta \)-Si\(_3\)N\(_4\) ceramics by field emission scanning electron microscopy (FE-SEM) [115, 116]. EBSD can be used to obtain a grain orientation map and pole figure map, thus, it provides a quantitative method for evaluating the texture in materials.

In addition to the above texture analysis methods, other methods, such as the Rietveld refinement and the XRD rocking-curve method [117], can also be applied to \( \beta \)-Si\(_3\)N\(_4\) ceramics [118, 119]. Each method has its own strengths and weaknesses. The choice of method, of course, depends on the ultimate purpose of the study. However, to develop a deep and quantitative understanding of the relationship between the texture and the anisotropic properties of materials. There are two major types of pole figure for determining the texture in \( \beta \)-Si\(_3\)N\(_4\) ceramics: \((hk0)\) and \((00l)\) pole figures, corresponding to the \( a \), \( b \)-axis and the \( c \)-axis orientations, respectively. Figure 7 shows typical (a) (200) and (b) (002) pole figures obtained from \( a \), \( b \)-axis and \( c \)-axis oriented \( \beta \)-Si\(_3\)N\(_4\) ceramics, which were produced from the plane-strain compression (PSC) [107] and tape-casting alignment (TCA) techniques [108], respectively.

The numbers in figure 7 represents multiples of a random distribution (mrd) of the grains obtained by comparing the orientation density on the projections between the untextured and textured samples; mrd = 1 and >1 for a random orientation and preferential orientation, respectively. The larger the maximum mrd value, the higher the degree of orientation. The maximum mrd values in the (200) and (002) pole figures 8 and 15, indicate the development of highly \( a \), \( b \)-axis [107, 109, 110] and \( c \)-axis oriented \( \beta \)-Si\(_3\)N\(_4\) ceramics [110, 111–113], respectively. Details of the (200) and (002) pole figure measurement can be found in [107] and [108], respectively.
in Si$_3$N$_4$ ceramics, a better and more accurate texture analysis is needed. Among the various methods discussed, the quantitative texture analysis techniques include pole figure analysis, EBSD, Rietveld refinement, and the rocking-curve method. Nevertheless, in several studies, it has been suggested that the XRD rocking-curve method is the most time-efficient and accurate texture analysis technique, as demonstrated for Al$_2$O$_3$ [118] and PMN-28 PT [119] ceramics.

However, the quantitative texture analysis of $\beta$-Si$_3$N$_4$ crystal in $\alpha$-Si$_3$N$_4$ green compacts is still difficult because of its limited amount (<10 wt%) and weak orientation, although the orientation, for example, induced by a strong magnetic field during slurry consolidation, can indeed be identified by the XRD peak intensity ratio. Quantitative texture analysis based on a polarized-light microscopy technique has been developed to evaluate the weak orientation in alumina green compacts [120, 121]. If $\beta$-Si$_3$N$_4$ exhibits optical anisotropy, this method can also be applied to evaluate its grain orientation in an $\alpha$-Si$_3$N$_4$ green compact.

5. Processing of textured Si$_3$N$_4$

In this article, the processing methods of textured Si$_3$N$_4$ ceramics have been classified into two types: hot-working (HW) and templated grain growth (TGG), as schematically illustrated in figure 8. In HW, the texture is primarily attributed to the rotation of elongated $\beta$-Si$_3$N$_4$ grains induced by a uniaxial stress imposed during sintering or forging. In TGG, the texture is attributed to the abnormal epitaxial growth of $\beta$-Si$_3$N$_4$ grains on initially oriented $\beta$-Si$_3$N$_4$ template (or seed) particles by the consumption of the fine-grained matrix during sintering. It is worth emphasizing that the term ‘templated grain growth’ was first proposed by Seabaugh et al [122] in 1997, and the authors defined this term as the use of oriented particles to obtain a texture by grain growth in the anisotropic direction of the template particles. Since then, TGG has been recognized as a term for a specific technique of fabricating textured ceramics. Both HW and TGG are thoroughly discussed in terms of their fabrication processes and the processing factors controlling the texture development.

5.1. Hot-working method

There are three different types of HW methods, including hot-pressing (HP), hot-forging (HF) and sinter-forging (SF).

5.1.1. Hot-pressing. Hot-pressing is a common densification method of Si$_3$N$_4$ ceramics that combines pressing and sintering processes while applying the uniaxial pressure. The starting material used for HP can be either loose powder or a compact form. This method enables the use of a smaller amount of sintering additives to obtain dense Si$_3$N$_4$ ceramics at a relatively lower temperature. At the same time, it promotes texture development in $\beta$-Si$_3$N$_4$. It has been reported that the texture in hot-pressed Si$_3$N$_4$ depends on the phase of the starting Si$_3$N$_4$ powder: $\alpha$ or $\beta$. In the case of raw $\alpha$-powder, the large elongated $\beta$-Si$_3$N$_4$ grains, preferentially developed by the $\alpha$–$\beta$ phase transformation, tend to be oriented with the c-axis (long axis) in the HP direction, thereby typically resulting in the a, b-axis oriented texture (figure 5) [8, 40, 53, 62, 65, 110, 123, 10].

Figure 8. Schematic illustration of texturing mechanisms of $\beta$-Si$_3$N$_4$ ceramics by (a) HW and (b) TGG method.
5.1.2. Hot-forging. In ceramics, HF is a superplastic forming technique, in which a uniaxial compressive or tensile stress is used to forge dense Si$_3$N$_4$ ceramics during heating. Dense Si$_3$N$_4$ ceramics can be fabricated by various sintering techniques, such as HP, pressureless sintering (PLS), and gas pressure sintering (GPS). In the case of tensile stress, the c-axis oriented texture preferentially develops. Wu and Chen [112] reported that the large tensile strains at 1550 °C allow hot-pressed β-Sialon with a fine-grained microstructure at 1550 °C to develop elongated grains and a very strong c-axis oriented texture. The degree of orientation increases with increased tensile strain. At the same time, they also observed a pronounced strain-induced grain growth phenomenon at a lower temperature, probably as a result of grain coalescence and grain unimpingement during deformation. However, in the case of compressive stress, the texture microstructure depends on the operating method of HF: simple compression (SC) (or axisymmetric forging) or PSC [111], as schematically shown in figure 9. SC tends to form the a, b-axis oriented texture [110, 111], but PSC tends to form the c-axis oriented texture [48, 61, 64, 107, 110, 111, 113].

5.1.3. Sinter-forging. SF is a novel technique that combines sintering and HF processes into a one-step process. In this technique, green compacts are produced by conventional cold-pressing and subsequent cold-isostatic pressing (CIP). Venkatachari and Raj [127] first developed this technique to obtain high-strength ceramics without large pore flaws, which were eliminated by the shear strain. They showed that, in Al$_2$O$_3$ ceramics, the strength began to increase at 20% strain and reached a maximum value at 60% strain. Kondo et al [51] used this technique to successfully develop textured Si$_3$N$_4$ with supermechanical properties. Using this technique they fabricated super-Si$_3$N$_4$ ceramics with a three-point bending strength of 2.1 GPa and a fracture toughness of 8.3 MPa m$^{1/2}$ in the direction perpendicular to the grain orientation, which is by far the highest reported value for strength. The super-Si$_3$N$_4$ was prepared by the following process: powder mixture (α-Si$_3$N$_4$–5 wt%Y$_2$O$_3$–3 wt%Al$_2$O$_3$) → cold-pressing (steel die: 40 mm length and 20 mm width) → CIP (490 MPa) → green compact (53% theoretical density (TD)) → SF (graphite channel die; 80 mm length and 20 mm width, pressure: 49 N, 1750 °C for 3 h in 0.1 MPa N$_2$) → c-axis oriented β-Si$_3$N$_4$. The heating profile used for SF and photographs of samples are shown in figure 10. Similar to HF, the texture also depends on the forging method: SC [128, 129] or PSC [51].

5.1.4. Grain rotation model. Assuming a rodlike morphology for β-Si$_3$N$_4$ grains and the absence of any interaction between neighboring grains, grain rotation models have been developed to describe the texture development in Si$_3$N$_4$ ceramics by Wu and Chen [112] and Lee and Bowman [111]. For grain rotation under tension, Wu and Chen obtained the following equation:

$$\ln \frac{\tan \phi_i}{\tan \phi} = -1.5 \varepsilon \ (\varepsilon > 0). \quad (15)$$

![Figure 9. Schematic illustration of superplastic deformation under (a) SC and (b) PSC.](image-url)
Figure 10. SF of Si$_3$N$_4$ ceramics: (a) heating profile and (b) view of CIP green (left) and sinter-forged (right) samples. During SF, the height shrinkage is $\sim 73\%$ based on the data provided by the authors (reproduced with permission from [51] ©1999 Blackwell Publishing Ltd).

Lee and Bowman described grain rotation under compression as follows.

Under SC,
\[
\ln \frac{\tan \phi_f}{\tan \phi_i} = \frac{3}{2} \frac{R^2 - 1}{R^2 + 1} (\varepsilon < 0),
\tag{16}
\]
\[
\frac{\sin \phi_i}{\sin \phi_f} = \frac{\tan \theta_i}{\tan \theta_f},
\tag{17}
\]

Under PSC,
\[
\ln \frac{\tan \phi_f}{\tan \phi_i} = \frac{2\varepsilon}{R^2 + 1} (\varepsilon < 0),
\tag{18}
\]
\[
\frac{\sin^2 \phi_i}{\sin^2 \phi_f} = \frac{\tan^2 \theta_i}{\tan^2 \theta_f},
\tag{19}
\]

In equations (15)–(19), $\phi$ and $\theta$ are the Euler angles describing the orientation of the grain lone axis with respect to a fixed 3D coordinate system, the subscripts $i$ and $f$, respectively, denote the initial and final states during deformation, $R$ is the aspect ratio of the grain, $\varepsilon$ is the tensile or compressive strain. Equations (15), (16) and (18) indicate that when $R$ becomes sufficiently large for the term $(R^2 - 1)/(R^2 + 1)$ to approach a constant of 1, the degree of orientation increases as the strain increases but independently of the aspect ratio. In this case, equation (15) is identical to equation (16).

Because most Si$_3$N$_4$ ceramics exhibit large values of $R$, the texture development in hot-forged Si$_3$N$_4$ basically depends on the strain, i.e., the texture development is essentially attributed to the grain rotation mechanism. This has been demonstrated both experimentally and theoretically by Lee and Bowman [111], as shown in figure 11. Nevertheless, the composition and amount of the additives have little effect on the degree of texture. Although the $a, b$-axis oriented texture can be enhanced by a larger height reduction, when the height reduction reaches a limit in the range of 57–69%, the degree of texture can no longer increase significantly because of the steric hindrance effect. Moreover, the authors demonstrated that the degree of $c$-axis texture increased significantly with increased height reduction during PSC. Xie et al [64, 107, 113] also demonstrated the dependence of texture development on strain during PSC within the true strain range of 0.30–0.95. The same authors further showed that when the aspect ratio was relatively small, for example, $R < 2.5$, it had a significant effect on the degree of texture in fine-grained $\beta$-Si$_3$N$_4$. As there is no difference in the mechanisms of the deformation and texture formation between HF and SF, the processing factors controlling the texture formation should be the same for both of them. Specifically, the degree of texture in sinter-forged Si$_3$N$_4$ should be governed by the forging strain. Table 3 lists typical examples of textured Si$_3$N$_4$ ceramics fabricated by HF and SF.
Table 3. Examples of textured Si$_3$N$_4$ fabricated by HF and SF.

| Authors            | Composition                                             | Condition | Phase | Density (% TD) | Condition | Eng. strain $\varepsilon$ (%)$^a$ | Density (% TD) | Texture type | Degree of texture                      |
|-------------------|---------------------------------------------------------|-----------|-------|----------------|-----------|-----------------------------------|----------------|-------------|----------------------------------------|
| Wu and Chen [112] | $\alpha$-Si$_3$N$_4$ + AlN + Al$_2$O$_3$ + Y$_2$O$_3$  | HP: 1550 °C | $\beta$-Sialon | > 85           | Tension: 1550 °C | 232                               | $c$-axis oriented | Pole figure analysis, Max. mrd = ∼3 $\parallel$ $c$-axis |
|                   |                                                         |           |       |                |           |                                   | $a, b$-axis oriented | Max. mrd = 4.3 $\perp$ $c$-axis |
|                   |                                                         |           |       |                |           |                                   | $c$-axis oriented | Pole figure analysis, Max. mrd = 6.9 $\parallel$ $c$-axis |
| Lee and Bowman [111] | $\alpha$-Si$_3$N$_4$ + 20 wt% YAG                     | PLS: 1750 °C | 100%  $\beta$-Si$_3$N$_4$ | > 85           | SC: 1750 °C | 69 $\rightarrow$ 94               | $c$-axis oriented | Plane $\perp$ Pressing: 33.4° |
|                   |                                                         |           |       |                |           |                                   | $a$-axis oriented | Plane $\parallel$ Extrusion: 24.1° |
|                   |                                                         |           |       |                |           |                                   |                |                          |
| Xie et al [113]   | $\beta$-Si$_3$N$_4$ + 7 wt% cordierite                | HP: 1750 °C | 100%  $\beta$-Si$_3$N$_4$ | > 99           | PSC: 1600 °C | 159                               | $c$-axis oriented | Orientation angle$^b$ |
|                   |                                                         |           |       |                |           |                                   | $a, b$-axis oriented | Plane $\perp$ Pressing: 33.4° |
|                   |                                                         |           |       |                |           |                                   | $c$-axis oriented | Plane $\parallel$ Extrusion: 24.1° |
| Kondo et al [61]  | $\alpha$-Si$_3$N$_4$ + 5 wt% Y$_2$O$_3$ + 2 wt% Al$_2$O$_3$ | GPS       | 100%  $\beta$-Si$_3$N$_4$ | > 99           | PSC: 1750 °C | 50 $\rightarrow$ 99               | $c$-axis oriented | Not reported |
|                   |                                                         |           |       |                |           |                                   | $a$-axis oriented | |
|                   |                                                         |           |       |                |           |                                   | $b$-axis oriented | |
|                   |                                                         |           |       |                |           |                                   | $I_{(101)}/I_{(210)} = 0.05$ | |
| Kondo et al [128] | $\alpha$-Si$_3$N$_4$ + 13 wt% Yb$_2$O$_3$ + 2 wt% SiO$_2$ | SC: 1900 °C | 100%  $\beta$-Si$_3$N$_4$ | > 97           | SC: 1900 °C | 60 $\rightarrow$ 97               | $c$-axis oriented | Plane $\perp$ Pressing$^c$ |
| Kondo et al [129] | $\alpha$-Si$_3$N$_4$ + 8 wt% Lu$_2$O$_3$ + 2 wt% SiO$_2$ | SC: 2000 °C | 100%  $\beta$-Si$_3$N$_4$ | > 97           | SC: 2000 °C | 60 $\rightarrow$ 97               | $c$-axis oriented | Plane $\perp$ Pressing$^c$ |
|                   |                                                         |           |       |                |           |                                   | $a, b$-axis oriented | $I_{(001)}/I_{(210)} = 0.035$ |

$^a$ Eng. strain $\varepsilon = (\Delta L/L_0) \times 100\%$ where $L_0$ is the initial dimension of the sample, $\Delta L$ is the dimension change during deformation. If only true strain, $\varepsilon$, is given in the literature, Eng. strain is calculated by the equation $\varepsilon = \ln(1 + \varepsilon)$.

$^b$ The orientation angle is the angle between the grain long axis and the extruding direction.

$^c$ $I_{(101)}/I_{(210)} = 1.06$ for the isotropic case (from PDF No. 33-1160 of $\beta$-Si$_3$N$_4$).
Surperplastic forming has been seen as a reliable and unique net-shape manufacturing technique of high-performance Si₃N₄ ceramics [130]. The constitutive equation for steady creep can generally be expressed as [131]

\[ \dot{\varepsilon} = \frac{A \sigma^n}{d^p}, \]  

(20)

where \( \dot{\varepsilon} \) is the strain rate, \( \sigma \) is the stress, \( d \) is the grain size, \( p \) and \( n \) are the grain-size and stress exponents, respectively, and \( A \) is a coefficient of proportionality. To achieve a larger superplastic deformation, a fine-grained microstructure with a tailored grain boundary phase is necessary through composition design and processing control. Therefore, the Si₃N₄ ceramics used for HF are normally sintered at relatively low temperatures to minimize the β-Si₃N₄ grain growth, as shown in table 3.

### 5.2. Templated grain growth method

#### 5.2.1. Template synthesis.

In the TGG method, the key point is the initial alignment of the single β-Si₃N₄ crystals, the so-called ‘seeds’ or ‘template’, during powder formation by various alignment techniques, such as cold-pressing alignment (CPA), extrusion alignment (EA), tape casting alignment (TCA) and SMFA, as summarized in table 4. In most cases, the rodlike morphology is the prerequisite for the alignment of β-Si₃N₄ crystals, specifically for EA and TCA. More importantly, a regulated rodlike morphology is also of particular importance to control the bimodal microstructure in self-reinforced Si₃N₄ ceramics. In EA and TCA, the β-Si₃N₄ seed particles must meet two major requirements: (i) single crystals must have a rodlike shape and (ii) the crystal size must be larger than that of the matrix of Si₃N₄ raw powder. In addition, an optimum aspect ratio as well as a narrow size distribution is also important. There are two major methods for synthesizing rodlike β-Si₃N₄ single crystals: powder sintering (PS) and combustion synthesis (CS).

#### a. PS method [132–134]

This method involves the heat treatment of an α-Si₃N₄ powder normally mixed with sintering additives at high temperatures, followed by grinding, sieving and a series of acid rinse treatments. During heating, the sintering additives provide a liquid phase to allow the α-phase to transform into the β-phase and grow anisotropically into rodlike β-Si₃N₄ crystals. The crystal morphology can be tailored by the type of α-Si₃N₄ powder, additives and heating conditions. Hirao et al. [132] developed a systematic method for synthesizing high-quality rodlike β-Si₃N₄ single crystals from a powder system of α-Si₃N₄, Y₂O₃ and SiO₂, in which crystals with average aspect ratios of 4 and 10 were obtained using molar ratios of Y₂O₃ : SiO₂ = 1 : 2 and 2 : 1, respectively. The mechanism underlying the effect of the Y₂O₃ : SiO₂ ratio on the β-Si₃N₄ crystal morphology was studied by Kitayama et al. [135], and they proposed that this effect is due to the difference in the solubility of Si₃N₄ caused by a change in the Y₂O₃ : SiO₂ ratio in the liquid phase. To obtain purer single β-Si₃N₄ crystals, Hirao et al. [132] developed an
5.2.2.1. CPA technique. Cold-pressing is most widely used for forming ceramic powders, usually using a metal die, particularly a stainless-steel die. During pressing, the shear strain tends to orient the anisotropic particles (e.g. platelets, ellipsoids, whiskers) by grain rotation, thereby resulting in texture development via TGG during sintering. Goto et al. [141] reported the formation of textured Si3N4 by CPA of an α-Si3N4 powder containing needle-like β-Si3N4 particles, followed by PLS. In their work, the needlelike β-Si3N4 particles were obtained by the heat treatment of α-Si3N4 powder with some Y2O3 and Al2O3 at 1700°C. The texture of Si3N4 is characterized by the c-axis of elongated β-Si3N4 grains being perpendicular to the cold-pressing direction, i.e. the a, b-axis oriented texture. Although this method has seldom been studied or

Table 5. Processes for obtaining rodlike β-Si3N4 seed crystals reported in the literature.

| Authors         | Method | Powder composition | Heating conditions | Seed characteristics^b |
|-----------------|--------|--------------------|--------------------|------------------------|
| Hirao et al [132] | PS     | α-Si3N4(SN-ES) + 5 mol%Y2O3 + 2.5 mol%SiO2 | 1850°C, 2 h, 0.5 MPa N2 | l = 2                   |
| Hirao et al [13] | PS     | α-Si3N4(SN-ES) + 5 mol%Y2O3 + 10 mol%SiO2 | 1850°C, 2 h, 0.5 MPa N2 | l = 4                   |
| Imamura et al [56] | PS   | α-Si3N4(SN-E10) + 5 mol%Y2O3 + 10 mol%SiO2 | 1850°C, 2 h, 0.5 MPa N2 | 0.47                    |
| Ramesh et al [133] | PS   | α-Si3N4(SN-E10) + 0.5 wt%Y2O3 | 1800°C, 0.35 atm N2 | 4.0                     |
| Chen et al [138] | CS    | Si + 48.8 wt% α-Si3N4 + 4.8 wt%Y2O3 | 5 MPa N2 | 0.5–2                 |
| Peng et al [140] | CS    | Si + 48.8 wt% α-Si3N4 + 2.4 wt%MgSiN2 | 3 MPa N2 | 10–15                  |

^a PS = powder sintering, CS = combustion synthesis.  
^b w = width, R is the aspect ratio, given by R = l/w (l = length).
used because of its poor texture control, it implies that the cold-pressing operation should be conducted carefully for Si$_3$N$_4$ powders containing rodlike $\beta$-Si$_3$N$_4$ particles to produce randomly oriented Si$_3$N$_4$ ceramics. Particular care is required in controlling the pressure parameter, because the orientation of rodlike $\beta$-Si$_3$N$_4$ particles is enhanced at a higher pressure.

5.2.2.2. EA technique. In this technique, the unidirectional alignment of rodlike $\beta$-Si$_3$N$_4$ seed particles can be achieved by extrusion through a die of the desired shape (e.g. square, circle) in an extrusion machine [50, 63, 142–144], as schematically illustrated in figure 13(a). The alignment is due to the grain rotation induced by shear strain. The grain rotation mechanism can be described by the model of Wu and Chen, i.e. the degree of orientation directly depends on the shear strain and aspect ratio of the seeds. Prior to the preparation of a ceramic paste for extrusion, a two-step mixing process is usually performed to achieve better homogenization between the seed particles, sintering additive and Si$_3$N$_4$ powder. In both steps an organic medium is preferably used, such as methanol or isopropyl alcohol. In the first step, intense milling, such as planetary milling, is used to mix the sintering additives and Si$_3$N$_4$ powder. The addition of seed particles follows the second step, and gentle milling should be used to avoid damage to the seed crystals, such as ball milling using nylon balls. To prepare the ceramic paste, water is normally used as a mixing medium, and other organic additives are added to improve the formability of ceramic powders. Table 6 lists the EA processes used for producing textured Si$_3$N$_4$ ceramics as reported in the literature. Two shapes are generally used for the extruded body: rods and rectangles. The stacking condition depends on the sintering method used. To achieve complete densification, PLS and GPS require CIP to increase the green density in comparison with HP.

5.2.2.3. TCA technique. Tape casting is a low-cost technique for the large-scale fabrication of ceramic substrates and bulk ceramics, in which a slurry is cast on a substrate through the action of a doctor blade that levels the slurry. It is also a powerful technique for producing a green sheet with oriented anisotropic ceramic particles, thereby developing textured ceramics, as schematically illustrated in figure 13(b). Tape casting can produce a green body with unidirectionally oriented rodlike $\beta$-Si$_3$N$_4$ seed particles. Figure 14 shows a typical tape-casting procedure for producing textured Si$_3$N$_4$ ceramics, in which a nonaqueous slurry is prepared by a
Table 6. Extrusion alignment processes for producing textured Si₃N₄ as reported in the literature.

| Authors          | β-Si₃N₄ seeds | Additives for aqueous ceramic paste | Morphology  | Amount   | Extruded body | Stacking method | Sintering (%) TD | Sintering | GD (%) SDb | Authors          | β-Si₃N₄ seeds | Additives for aqueous ceramic paste | Morphology  | Amount   | Extruded body | Stacking method | Sintering (%) TD | Sintering | GD (%) SDb |
|------------------|---------------|-----------------------------------|-------------|----------|---------------|----------------|------------------|------------|-------------|-----------------|---------------|-----------------------------------|-------------|----------|---------------|----------------|------------------|------------|-------------|
| Muscat et al     | ∼53 PLS, 95% C, 1 h | Hydroxy-propyl cellulose (HPC) 60 mm | Rectangle   | 0.1–0.3 µm, R = 4 | Binder: YB-113G and YB-113C Rod | CIP: 3-400 MPa | ∼3 | 0.1–0.3 µm, R = 4 | Zou et al       | Whisker (SN-WB) | Red | Cold pressing: 40 MPa | Rectangle 50 mm | 3 wt% Binder: PEG 1000 | 1750 C, 2 h, 0.1 M Pa N₂ |
| Teshima et al    | ∼54 GPS, 98% C, 2 h | Binder: YB-113G and YB-113C Rod | Wire 0.2 mm | 0.44 µm, R = 8 | Plasticizer: glycerine 1820 °C Lubricant: paraffin 0.1 M Pa N₂ | Cold pressing: 1 MPa | ∼54 | 0.44 µm, R = 8 | Belmonte et al | Whisker (SN-WB) | Red | Cold pressing: 1 MPa | Rectangle 50 mm | 5 wt% Binder: PEG 1000 | 1750 C, 2 h, 0.9 M Pa N₂ |
| Nakamura et al   | ∼54 GPS, 98% C, 2 h | Binder: YB-113G and YB-113C Rod | Wire 0.2 mm | 0.44 µm, R = 8 | Plasticizer: glycerine 1820 °C Lubricant: paraffin 0.1 M Pa N₂ | Cold pressing: 1 MPa | ∼54 | 0.44 µm, R = 8 | Zou et al       | Whisker (SN-WB) | Red | Cold pressing: 40 MPa | Rectangle 50 mm | 3 wt% Binder: PEG 1000 | 1750 C, 2 h, 0.1 M Pa N₂ |

Three-step milling process [43]. The initial planetary milling is used to homogenize the Si₃N₄ powder and sintering additives, which is followed by a two-step ball milling process. Note that various additives are needed for the preparation of the tape-casting slurry. The role of the dispersant is to ensure the stability of the slurry, and the roles of the binder and plasticizer are to ensure adequate strength and flexibility of the tape. Details of the tape-casting technology for producing ceramics can be found in several review articles [145–147]. Two major nonaqueous tape-casting processes have been developed to produce textured Si₃N₄, as listed in Table 7.

Recently, because of the safety, economic and environmental considerations, the aqueous tape-casting of Si₃N₄ ceramics has received attention [148–150], as listed in Table 8. Although these studies do not involve the use of β-Si₃N₄ seeds, they will undoubtedly be of great help in producing textured Si₃N₄ by the aqueous TCA technique. Similarly to EA, the slurry incorporated with β-Si₃N₄ seeds should be gently mixed, typically by ball milling using nylon balls. Also, CIP treatment is required for PLS and GPS to achieve complete densification.

a. Whisker orientation model

A model for whisker orientation during tape casting has been proposed by Wu and Messing [151], described as follows

\[
\frac{d\alpha}{d\tau} = \frac{1}{4} \frac{R^2 - 1}{R^2 + 1} G \sin 2\alpha \sin 2\beta, \quad (21)
\]

\[
\frac{d\beta}{d\tau} = G \left( 1 + \frac{R^2 - 1}{R^2 + 1} \cos 2\beta \right), \quad (22)
\]

where \( R \) is the aspect ratio of a whisker, or more generally, any shaped particle (e.g., platelet), \( G \) is the shear rate during tape casting, and \( \alpha \) and \( \beta \) are the angles determining the position of the whisker in a polar coordinate system, as shown in Figure 15. In addition, if \( \phi \) is defined as the angle between the tape-casting direction and the whisker orientation, the relationship between \( \phi, \alpha \) and \( \beta \) is

\[
\sin \phi = \sqrt{\cos^2 \alpha + \cos^2 \beta \sin^2 \alpha}. \quad (23)
\]

The perfect orientation requires \( \phi = 0^\circ \) or \( \alpha = \beta = 90^\circ \). Equations (21) and (22) predict that the whisker orientation is governed by the following factors

(i) Shear rate

For a Newtonian suspension, the shear rate \( G \) is directly proportional to the tape-casting rate and inversely proportional to the tape thickness. Therefore, the rates of whisker orientation can be increased by increasing the shear rate (e.g., casting rate) or by decreasing the blade opening during casting.

(ii) Aspect ratio

As \( R \) becomes large, the term \( (R^2 - 1)/(R^2 + 1) \) approaches 1. Thus, longer whiskers undergo a faster rate of orientation. Moreover, if \( R \) is greater than 10, these longer whiskers should behave in a similar manner assuming that the whisker content in the slurry is below the percolation threshold and that whiskers do not interact.
Figure 14. Processing route for TGG using TCA and GPS.

with one another or with the particulate phase during tape casting.

(iii) Initial position
The initial position of a whisker during tape casting also affects the orientation. There are three different initial positions as follows.

**Pure rotation:** if the whisker lies in the \( LZ \) plane, it only rotates about the \( T \)-axis, thereby it is aligned in the \( L \)-axis or casting direction.

**Pure spinning:** if the whisker lies in the \( T \)-axis, it only spins about its own axis and cannot be aligned in the \( L \)-axis for a 1D velocity gradient.

**Mixed mode:** during tape casting, the whisker alignment occurs as a result of both rotation and spinning. As \( \beta \) approaches 90°, the net torque exerted by the fluid on the whisker approaches zero.

However, the factors controlling the shear rate \( G \) are not considered in Wu and Messing’s model. During tape casting, the flow behavior of a Newtonian fluid is due to a combination of pressure and viscous forces. To describe the flow characteristics of the slurry during tape casting, Kim et al. [152] proposed a parameter \( \Pi \) to define the ratio of
| Table 7. Nonaqueous TCA processing of textured Si$_3$N$_4$ by GPS as reported in the literature. |
|---|---|---|---|---|---|---|---|
| Authors | Solvent | Dispersant | Binder | Plasticizer | Stacking condition | Green body |
| | | | | | | |
| Hirao et al [43] | Toluene/ n-butanol = 4/1 | Diamine | Polyvinyl butyral (PVB) | Dioctyl adipate | 130 °C, 6 h, >99 | 1850 °C, 6 h, >99 |
| Park and Kim [153] | Methylisobutyl ketone (MIBK) | PRT | Polyvinyl butyral (PVB) | Dibutyl phthalate | 80 °C, 4 h, >99 | 1875 °C, 4 h, >99 |

| Table 8. Aqueous tape casting of Si$_3$N$_4$ ceramics as reported in the literature. |
|---|---|---|---|---|---|---|
| Authors | Sintering additives | Dispersant | Binder | Plasticizer | Others | Slurry conditions |
| | | | | | | |
| Bitterlich et al [148, 149] | Y$_2$O$_3$ + Al$_2$O$_3$ | Dolapix A88: PC 33 = 90 : 1 (in weight) | Copolymer (Mowilith DM 765) | Defoamer: Fatty alcohol | 10 | No stacking, Single tape for sintering Pressure: 14 MPa, GPS: 1800 °C, 1 h, 5 MPaN$_2$, ~98 [148] |
| Zhang et al [150] | Y$_2$O$_3$ + Al$_2$O$_3$ | PEI + Citric acid ammonium salt | PVA | Glycerol | 9.3 | Simple stacking in a graphite die GPS: 1800 °C—30 min SPS: 1600 °C—3 min 

>95 [149]
pressure force to viscous force as follows:

\[ \Pi = \frac{\Delta P H^2}{2\eta L U}, \]

\[ \Delta P = \rho g H, \]

where \( \Delta P \) is the pressure exerted by the slurry head, \( H \) is the blade gap used, \( \eta \) is the viscosity of the slurry, \( L \) is the length of the doctor blade, \( U \) is the casting velocity, \( \rho \) is the density of the slurry and \( H \) is the height of the slurry in the reservoir. The viscous force determines the shear behavior necessary for particle rotation and alignment. Equation (24) indicates that the degree of whisker orientation can be increased by increasing the casting velocity (\( U \)), slurry viscosity (\( \eta \)) and whisker fraction.

**b. Modified TCA technique**

As discussed above, the whisker orientation strongly depends on the casting velocity, i.e. the velocity gradient along a single whisker. As reported previously, the length of rodlike \( \beta-Si_3N_4 \) seed crystals is usually in the order of \( \sim 10 \mu m \), which is smaller than \( 1/10^2 \) of the width of the flow during tape casting, thereby resulting in a lower velocity gradient in the local region. In this case, using the method of only adjusting tape-casting velocity, it is very difficult to achieve a highly unidirectional alignment of seed particles by conventional tape casting, particularly for a low content of seeds, as shown in figure 13(b). This is the reason why the TCA technique does not normally produce as highly \( \beta-Si_3N_4 \) whiskers as the EA technique. To solve this problem, Park and Kim [153] developed a modified TCA, in which an array of sharp guides (e.g. pins) are set at the exit of the reservoir to divide the flow into \( n \) narrow flows, thereby increasing the torque for aligning the whiskers by a factor of \( n^2 \), as schematically illustrated in figure 16. Using this method, they produced highly \( c \)-axis oriented \( \beta-Si_3N_4 \) with 3 wt% \( \beta-Si_3N_4 \) whiskers (UBE SN-WB), using an array of sharpened pins 0.7 mm apart from each other, a casting speed of 10 mm s\(^{-1}\) and a blade gap of 0.45 mm.

**5.2.2.4. SMFA technique.** Recently, SMFA has received more attention for the fabrication of textured nonmagnetic ceramics, such as Al\(_2\)O\(_3\), TiO\(_2\), AlN, ZnO, hydroxyapatite (HAp). When a strong magnetic field (typically \( \geq 10 \) T) is imposed, the nonmagnetic ceramic particles are oriented during slurry consolidation, thereby resulting in the formation of textured ceramics during sintering. Important developments have been discussed in a review article by Sakka and Suzuki [154]. Compared with the other three techniques, SMFA has outstanding advantages of the independence of grain morphology and its applicability to even very fine nanoscale grains. Another attraction is the versatility and simplicity of the fabrication process, because various colloidal forming approaches are well developed, such as slip casting [155], electrophoretic deposition [156] and gel casting [157].

**a. Magnetic field alignment theory**

(i) **Rotating condition**

When a single crystal with magnetic anisotropy is placed in a magnetic field, it will rotate if the anisotropic magnetic energy is higher than the energy of the thermal motion. The condition for rotation is expressed as [158]

\[ V > \frac{2\mu_0 k_B T}{|\Delta \chi| B^2}, \]

where \( \Delta \chi = \chi_a - \chi_c \), is the anisotropy of the magnetic susceptibilities, \( V \) is the volume of the material, \( B \) is the applied magnetic field, \( \mu_0 \) is the permeability in a vacuum, \( k_B \) is the Boltzmann constant and \( T \) is the absolute temperature.

(ii) **Static magnetic field alignment model**

If a crystal is dispersed in a viscous liquid of viscosity \( \eta \), the crystal rotation in a static magnetic field (SMF) can be described as [158]

\[ \ln \frac{\tan \xi_f}{\tan \xi_i} = -\frac{t}{\tau} \quad (0 < \xi < 90^\circ), \]

where \( \xi \) is the angle between the magnetic field direction and the crystal axis of the largest magnetic susceptibility. The subscripts ‘\( i \)’ and ‘\( f \)’, respectively, denote the initial and final states during alignment, \( t \) is the time and \( \tau^{-1} \) is defined as the alignment rate

\[ \tau^{-1} = \frac{V|\Delta \chi| B^2}{L \mu_0}, \]
Rotating magnetic field alignment model

Figure 17. Polar coordinate system describing the whisker orientation in an RMF on the XY plane with an angular velocity of $\omega$. $\xi$ is the angle between the magnetic field direction and the whisker long axis.

where $L$ is a hydrodynamic term related to the particle shape. For a spherical particle of radius $d$, equations (26) and (28) can be rewritten as

$$d > \left( \frac{3k_B\mu_0 T}{2\pi |\Delta\chi| B^2} \right)^{1/3}$$  \hfill (29)

$$\tau^{-1} = \frac{|\Delta\chi| B^2}{6\eta\mu_0}.$$  \hfill (30)

For an ellipsoidal (or prolate) crystal with short-axis radius $d$ and aspect ratio $R$, equations (26) and (28) can be rewritten as

$$d > \left( \frac{3k_B\mu_0 T}{2\pi |\Delta\chi| B^2 R^2} \right)^{1/3},$$  \hfill (31)

$$\tau^{-1} = F(R) \frac{|\Delta\chi| B^2}{6\eta\mu_0},$$  \hfill (32)

$$F(R) = \frac{3R \left(1 - 2R^2\right)\ln \left( \frac{8\sqrt{R^2 - 1}}{8\sqrt{R^2 - 1}} \right) - 2R\sqrt{R^2 - 1}}{4 \left(R^2 - 1\right) \left(R^2 + 1\right) \sqrt{R^2 - 1}}.$$  \hfill (33)

If $R \rightarrow 1$, then $F(R) = 1$ and equation (30) is identical to equation (32); if $R \rightarrow \infty$, then $F(R) = -3/2$ and equation (32) describes the case of a whisker (or fiber).

(iii) Rotating magnetic field alignment model

In an SMF, the crystal is aligned with the largest magnetic susceptibility axis parallel to the magnetic field. If $\chi_c > \chi_{a,b}$, the crystal is aligned with the $c$-axis parallel to the magnetic field, namely, the uniaxial or unidirectional alignment; if $\chi_c < \chi_{a,b}$, the crystal is aligned with the $a$, or $b$-axis parallel to the magnetic field, namely, the planar alignment. In the case of $\chi_c < \chi_{a,b}$, a rotating magnetic field (RMF) can be used to achieve the alignment of the $c$-axis. As schematically shown in figure 17, the crystal alignment model in an RMF (B) can be described as follows [159]

$$\frac{d\theta}{dr} = (2\tau)^{-1} \cos^2 (\phi - \omega t) \sin 2\theta,$$  \hfill (34)

where $\omega$ is the angular velocity of the rotation of the magnetic field on the XY plane. The alignment rate $\tau^{-1}$ is given by equations (30) and (32) for a spherical and ellipsoidal crystal, respectively.

Equations (29)–(32), (34) and (35) indicate the conditions necessary for the magnetic alignment of crystals during slurry consolidation, including the following.

(a) The particle should be anisotropically magnetic, normally, with a noncubic crystal structure (e.g. hexagonal, tetragonal).

(b) The size of the particle is sufficiently large to overcome the energy of thermal motion.

(c) The suspension should be well deagglomerated or dispersed and exhibit low viscosity, because only single crystals can be aligned in the magnetic field.

(d) The rotating velocity of the magnetic field should be sufficiently high to align along the $c$-axis. Note that the RMF is achieved by rotating the sample in the SMF, where the magnetic field is parallel to the rotating plane [159]. In this case, the rotating velocity corresponds to the rotating velocity of the sample.

Figure 18 shows the dependence of the minimum particle size necessary for the rotation of a spherical particle on the magnetic field (T), calculated by equation (29).

$$\frac{d\phi}{dr} = -(2\tau)^{-1} \sin (\phi - \omega t),$$  \hfill (35)

where $\phi$ is the angular velocity of the rotation of the magnetic field on the XY plane. The alignment rate $\tau^{-1}$ is given by equations (30) and (32) for a spherical and ellipsoidal crystal, respectively.

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Figure 18 shows the dependence of the minimum particle size necessary for the rotation of a spherical particle on the magnetic field (T), calculated by equation (29).
Figure 19. Schematic drawings of orientation mechanisms of $\beta$-Si$_3$N$_4$ grains during slip casting in an (a) SMF and (b) RMF.

feature of SMFA in comparison with the other three alignment techniques.

b. SMFA of Si$_3$N$_4$

Because of their hexagonal structure, a strong magnetic field of $\geq 10$ T leads to the alignment of both $\alpha$- and $\beta$-Si$_3$N$_4$ crystals during the slip casting of $\alpha$-Si$_3$N$_4$ powders. However, the SMF leads only to the alignment of both $\alpha$- and $\beta$-Si$_3$N$_4$ crystals with the $a$ or $b$-axis parallel to the magnetic field, or the planar orientation because of the magnetic susceptibility of $\chi_{a,b} > \chi_c$ [163], as schematically illustrated in figure 19(a). Hence, the $a$, $b$-axis oriented $\beta$-Si$_3$N$_4$ is preferentially developed by sintering, during which the $\alpha$-phase completely transforms into the $\beta$-phase [103, 163]. To align the $c$-axis of $\beta$-Si$_3$N$_4$ crystals, an RMF has been used for the slip casting of $\alpha$-Si$_3$N$_4$ powders that contain a small fraction of $\beta$-Si$_3$N$_4$ crystals [82, 164, 165], as shown in figure 19(b). In the RMF, slip casting is perpendicular to the magnetic field direction, and the rotating speed of the sample should be optimized to achieve the highest orientation. It was found that a rotating speed of 10 rpm is sufficient to achieve greater orientation of the $c$-axis of $\beta$-Si$_3$N$_4$ single crystals [82].

The magnetic field alignment requires a suspension with well-deagglomerated $\beta$-Si$_3$N$_4$ particles. Slurry consolidation also requires a well-stabilized suspension for developing a high-quality green body, better sinterability and high-quality sintered products. In SMFA, the dispersion and stability of the Si$_3$N$_4$ suspension (aqueous or nonaqueous) are crucial for controlling the texture development and properties of $\beta$-Si$_3$N$_4$ ceramics. Because of the need for safety, aqueous processing is preferable for producing Si$_3$N$_4$ ceramics. The most common polyelectrolytes used for stabilizing aqueous Si$_3$N$_4$ suspensions are acrylic-based polymers, such as poly(acrylic acid) (PAA) or its ammonium salt (PAA-NH$_4$), known as anionic polyelectrolytes [166–168]. Recent studies have shown that polyethylenimine (PEI) is an effective cationic dispersant for aqueous Si$_3$N$_4$ suspension, because it can provide better stabilization of the suspension by maintaining the suspension pH and producing a steric or electrosteric effect [169–171].

For the SMFA of $\alpha$-Si$_3$N$_4$ powders, previous studies have revealed one important fact: the $\beta$-Si$_3$N$_4$-phase crystal exhibits stronger orientation ability than the $\alpha$-Si$_3$N$_4$ phase crystal despite the minor phase, for example, $\sim 1$ and $\sim 5$ wt% for SN-E5 and SN-E10 powders, respectively [163]. One reason for this is that, compared with $\alpha$-Si$_3$N$_4$ particles with an equiaxed shape, $\beta$-Si$_3$N$_4$ particles are less agglomerated and have a predominantly rodlike shape. The other reason is probably because the $\beta$-Si$_3$N$_4$ crystal exhibits a larger value of the anisotropic magnetic susceptibility than the $\alpha$-Si$_3$N$_4$ crystal according to equation (26). Furthermore, as the $\alpha \rightarrow \beta$ phase transformation occurs via the solution-reprecipitation process, the solution probably disturbs the crystallographic orientation of the $\alpha$-phase. However, due to the greater stability, the oriented $\beta$-Si$_3$N$_4$ grains act as seeds to promote the development of oriented large elongated $\beta$-Si$_3$N$_4$ grains via anisotropic grain growth after and during the phase transformation, namely, the TGG mechanism. Although highly $a$, $b$-axis oriented $\beta$-Si$_3$N$_4$ can be produced via SMFA by the slip casting of $\alpha$-Si$_3$N$_4$ powder (SN-E10) without additional $\beta$-Si$_3$N$_4$ seed particles, the use of $\beta$-Si$_3$N$_4$ seeds with tailored morphology is more efficient for not only increasing the degree of orientation, particularly the $c$-axis orientation, but also improving the bimodal microstructure.

The authors studied the effect of Si$_3$N$_4$ powder type ($\alpha$ or $\beta$-Si$_3$N$_4$) on texturing $\beta$-Sialon by an SMF and reaction.
Figure 20. Schematic illustrations of sintering shrinkage behavior of textured Si₃N₄ by (a) EA, (b) TCA, (c) slip casting in an SMF and (d) slip casting in an RMF. The left figure is the schematic texture and the right figure is the shrinkage curve as a function of time (t). t₀ is the time when the shrinkage begins.

PLS [172] and clearly demonstrated that β-Si₃N₄ crystals exhibit substantially stronger orientation than α-Si₃N₄ crystals. The β-raw powder produced a highly a, b-axis oriented β-Si₃N₄ green body with a Lotgering orientation factor of up to 0.97. During sintering, the β-raw powder allowed the a, b-axis oriented β-Sialon to retain the Lotgering orientation factor similar to and even higher than that of β-Si₃N₄ in the green body. In contrast, the α-raw powder led to a faster transformation rate of α/β-Si₃N₄ to β-Sialon but a substantially lower texture in β-Sialon. Using the slip casting of Si powder containing β-Si₃N₄ particles in an SMF, the authors also obtained a, b-axis aligned sintered reaction-bonded silicon nitride (SRBSN) [100]. Because of the cheap Si raw powder, the 60% weight gains by silicon nitridation, the machinability of RBSN body and the smaller sintering shrinkage, SRBSN is a well-known cost-effective Si₃N₄ ceramic material, thus, this work suggested that the combination of SMFA and SRBSN is a promising method for fabricating cost-effective textured Si₃N₄ as well as Sialon.

5.2.3. Anisotropic sintering shrinkage. Shrinkage anisotropy is an important phenomenon that occurs during the sintering of Si₃N₄ compact-containing oriented β-Si₃N₄ seed particles. The anisotropy is primarily associated with the texture development via the preferential grain growth of elongated β-Si₃N₄ seeds. The shrinkage anisotropy behavior in 3D space depends on the processing route and texture structure, as schematically illustrated in figure 20. The shrinkage is defined as

\[ \varepsilon = \frac{L - L_0}{L_0} \times 100\% \]  

where L₀ and L are the sample dimensions before and after sintering, respectively. Furthermore, the shrinkage anisotropy factor, fₛₐ, is defined as

\[ f_{SA} = \left( 1 - \frac{\varepsilon_\parallel}{\varepsilon_\perp} \right) \times 100\% \]  

where \( \varepsilon_\parallel \) and \( \varepsilon_\perp \) are the shrinkages in the directions parallel and perpendicular to the c-axis of elongated β-Si₃N₄ grains, respectively. Muscat et al [142] reported an \( f_{SA} \) value of 75% for extruded Si₃N₄ with β-Si₃N₄ whiskers. Teshima et al [50] reported an \( f_{SA} \) value of 67% for extruded Si₃N₄ with rodlike β-Si₃N₄ seeds. Using a modified TCA technique, Bae et al [108] reported an \( f_{SA} \) value of 70% for textured Si₃N₄ with fine β-Si₃N₄ whiskers, which is higher than that (≈52%) for tape-cast Si₃N₄ with coarser β-Si₃N₄ whiskers.

Generally, the \( f_{SA} \) value increases with the increased degree of orientation by increasing the degree of orientation of seed particles and the seed content. Figure 21 illustrates the increased shrinkage anisotropy with increased β-Si₃N₄
whisker content obtained by modified TCA [173]. On the basis of sintering shrinkage anisotropy, Park and Kim [173] proposed a model to quantify the degree of orientation of the large elongated grains in c-axis aligned Si₃N₄ developed from β-Si₃N₄ whisker templates. Figure 22 gives the shrinkage anisotropy in textured Si₃N₄ prepared by SMFA using slip casting [82, 163]. It is clear that texture development results in substantially higher shrinkage anisotropy in comparison with an untextured sample. Nevertheless, the addition of β-Si₃N₄ seeds does not seem to contribute to the shrinkage anisotropy, which is different from the report of Part and Kim [173]. This is most likely to be due to the morphology of the β-Si₃N₄ seeds used. If the β-Si₃N₄ seeds exhibit a whiskerlike shape and a larger size than the matrix grains, the shrinkage anisotropy should be enhanced [174]. Despite this, the shrinkage anisotropy can be seen as a direct indicator of texture development in Si₃N₄ ceramics.

5.2.4. Factors affecting texture development.

a. β-Si₃N₄ seed morphology and content

Texture development via TGG is determined by the alignment of seed particles, which is controlled by the β-Si₃N₄ seed morphology and content. The dependence of texture development on β-Si₃N₄ seed morphology is related to the seed alignment technique. For EA and TCA, the β-Si₃N₄ seeds must be externally added because of the requirements of a rodlike morphology and a larger size than the matrix grains. In both EA and TCA, the texture development can be enhanced by increasing the aspect ratio and the β-Si₃N₄ seed content. Imamura et al [56] reported that texture development in β-Si₃N₄ is enhanced by increasing the aspect ratio and decreasing the diameter. The larger aspect ratio results in a higher degree of orientation, as the Wu and Messing model predicts [151]. When the aspect ratio remains constant, the seed density in the matrix decreases as the diameter increases, i.e., the number of oriented seed particles decreases as the diameter increases, thereby increasing the degree of orientation of large elongated grains during sintering. Bae et al [108] reported that for tape casting, fine whiskers with a diameter of 1.1 µm and an aspect ratio of 11.1 result in a higher degree of texture in sintered Si₃N₄ than coarse whiskers with a diameter of 2.8 µm and an aspect ratio of 6.7. The enhanced texture development is due to a combination of the larger aspect ratio and the larger seed density in the finer whiskers. Furthermore, the degree of texture increases with increased seed density regardless of the seed size. Also, for a given β-Si₃N₄ seed, the degree of texture increases as the seed content increases [75, 175].

Although SMFA is independent of the rodlike morphology, it strongly depends on the β-Si₃N₄ seed content. Again, β-Si₃N₄ seeds with size larger than the α-Si₃N₄ matrix should be beneficial for enhancing the texture via TGG during sintering. According to the Dressler model [94], if the size of β-nuclei is below the critical size, they will dissolve and reprecipitate on the surrounding large β-Si₃N₄ grains. This means that, to guarantee the stability of oriented β-nuclei during sintering, the β-nuclei should be sufficiently large. Nevertheless, because of the high solubility, α-grains should dissolve more rapidly than β-grains with the same size. This suggests that the β-nuclei may have the same order of grain size as the matrix α-Si₃N₄ particles used for the SMFA technique. This has been clearly demonstrated in the literature [82, 103, 163–165]. Although the texture can develop from the preexisting β-Si₃N₄ nuclei in the α-Si₃N₄ raw powder, the number of the preexisting β-Si₃N₄ nuclei is usually insufficient to achieve a high degree of texture in β-Si₃N₄, particularly for the c-axis texture generated by an RMF. Table 9 gives the effects of α-Si₃N₄ powder and β-Si₃N₄ seeds on the degree of orientation of β-Si₃N₄ grains in green bodies by slip casting using an SMF and RMF. It is evident that the addition of β-seeds markedly increases

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**Figure 21.** Effect of β-Si₃N₄ whisker content on the shrinkage anisotropy of Si₃N₄ compacts prepared by modified TCA during GPS. The authors only present the data of α, and β, in their work. If α were given, the f_SA values would be larger.

**Figure 22.** Shrinkage anisotropy in textured Si₃N₄ prepared by SMFA using slip casting. Sample TE10S contains 5 wt% equiaxed β-Si₃N₄ seeds relative to sample TE10, and E10 means the raw UBE SN-E10 powder. Sample UTE10 was prepared in the absence of a magnetic field. For the RMF of 12 T, the rotating speed of the sample was fixed at 10 rpm.

\[ f_{SA} = \left(1 - \frac{\alpha}{\beta}\right) \times 100\% \]
Table 9. Degree of orientation of β-Si₃N₄ grains in green bodies by slip casting using an SMF and RMF [82,163].

| Sample | XRD intensity ratio | Logg (I(002)/I(200)) |
|--------|---------------------|----------------------|
|        | SMF (I(200)/I(101)) | RMF (I(002)/I(200))  |
| TE10   | TS (⊥ B) 2.90       | SS (∥ B) 0.75        |
|        | ≈0.5               | ≈0                  |
| TE10S  | 12.4               | 0.89                |
|        | 5.07               | ≈0                  |
| TE5    | 1.67               | 1.43                |
| TE5S   | 13.43              | 1.08                |

TS = top surface, perpendicular to the magnetic field for SMF but parallel to the magnetic field for RMF; SS = side surface, parallel to the magnetic field for SMF but perpendicular to the magnetic field for RMF; see figure 19.

b. Sintering conditions

Sintering is responsible for most of the texture development in ceramics via TGG. During sintering, the oriented β-Si₃N₄ seeds grow both anisotropically and abnormally by consuming the surrounding small matrix grains by the α–β phase transformation and subsequent grain growth by Ostwald ripening, eventually resulting in the enhanced texture. For EA and TCA, the addition of large β-Si₃N₄ seeds tends to hinder the densification; thus, the GPS method at a higher temperature (>1800 °C) is preferable, particularly for the case of low amounts of sintering additives, as shown in tables 6 and 7. The higher temperature and prolonged sintering favors the grain growth and thus the texture development. However, the degree of texture eventually reaches a limit as the sintering time increases, because the grain growth is hindered by the impingement of neighboring β-Si₃N₄ grains [45, 75, 163]. As shown in figure 23, the use of β-seeds is more efficient for developing highly textured Si₃N₄ via SMFA, particularly c-axis textured Si₃N₄ by an RMF, compared with prolonged sintering. As shown in figure 24, the addition of β-seeds increases the number of initially oriented β-Si₃N₄ nuclei and promotes the growth of large elongated grains, thereby increasing the degree of texture. The core-rim structure (indicated by white arrows) implies that the oriented large elongated grains grow epitaxially from the initially oriented β-Si₃N₄ particles during slip casting. Although some of the β-Si₃N₄ seeds do not have a rodlike morphology, the elongated grain growth enables them to grow into large elongated grains. In addition, Park et al [176] also observed some large elongated grains with more than one core grown using a Y₂O₃–Al₂O₃ additive composition, probably as a result of the coalescence of grains growing from separate β-Si₃N₄ seeds.

c. Seed alignment technique and processing parameters

Because the c-axis oriented β-Si₃N₄ can have the maximum anisotropy in its properties, the alignment of the c-axis of β-seeds is of great interest. Compared with TCA and SMFA, EA is more effective for achieving this goal. However, the high orientation of the c-axis can also be achieved by modified TCA (figure 16) through tailoring the slurry properties and casting parameters. Furthermore, the orientation of the c-axis can also be achieved by an RMF. Although no data is available concerning the effect of slurry properties on the orientation of the β-seed nuclei via SMFA, it should be emphasized that a well-dispersed slurry is a prerequisite for this alignment technique, as discussed by Sakka and Suzuki in their review article [154]. Therefore, the key points for developing highly textured β-Si₃N₄ are as follows:

(i) High orientation of the β-Si₃N₄ seed crystals,
(ii) Optimum content of β-Si₃N₄ seed,
(iii) High-temperature sintering/annealing.
Table 10. Vickers hardness (HV) of untextured and textured Si$_3$N$_4$ as reported in the literature.

| Author          | Material                  | HV (GPa) || c-axis | ⊥ c-axis |$f_{\text{SHA}}$ (%) | Schematic of hardness test |
|-----------------|---------------------------|----------|-------|---------|---------------------|---------------------------|
| Goto et al [141]| a, b-axis aligned         | 15.7     | 14    | 11      |                     |                           |
| Santos et al [126]| Weakly a, b-axis aligned | 16.3     | 15.9  | 3       |                     |                           |
| Belmonte et al [144]| c-axis aligned   | 15.2     | 14.3  | 6       |                     |                           |
| Teshima et al [50]| Highly c-axis aligned    | 15.8     | 13.4  | 15      |                     |                           |
| Nakamura et al [63]| Highly c-axis aligned  | 17       | 12.6  | 26      |                     |                           |

$\text{f}_{\text{SHA}} = (1 - \frac{\text{HV}_{\perp}}{\text{HV}_{\parallel}}) \times 100\%$

6. Anisotropic properties of textured Si$_3$N$_4$

6.1. Hardness

In addition to the porosity, grain boundary phase and grain size [177], the orientation of elongated β-Si$_3$N$_4$ grains also affects the hardness of Si$_3$N$_4$ ceramics [50, 63, 126, 141, 144], as clearly demonstrated by the experimental data listed in table 10. The plane parallel to the grain alignment (or c-axis) exhibits greater hardness than the plane perpendicular to the grain alignment. Thus, the texture allows the textured material to possess greater hardness in the plane parallel to the grain alignment than the untextured material prepared under the same conditions. The hardness anisotropy depends on the degree of orientation. Higher orientation of the c-axis of elongated grains results in higher hardness anisotropy. The anisotropy in textured Si$_3$N$_4$ is essentially due to the intrinsic anisotropy in the hardness of β-Si$_3$N$_4$ single crystals [178–180], as shown in table 11. The intrinsic hardness anisotropy reaches ~46% for the β-Si$_3$N$_4$ single crystal. The intrinsic hardness anisotropy of the single crystal is due to the difference in the number of bonds per unit area on different crystallographic planes. As Chakraborty and Mukerji [178] proposed, the intrinsic hardness anisotropy of a single crystal, $f_{\text{IHA}}$, can be expressed in theory as

$$f_{\text{IHA}} = (1 - \frac{\text{HV}_{\perp}}{\text{HV}_{\parallel}}) \times 100\% = (1 - \frac{100c}{63a}) \times 100\%,$$

(38)

where $a$ and $c$ are the unit cell parameters. For β-Si$_3$N$_4$, $a = 0.7608$ and $c = 0.2911$ nm, and $f_{\text{IHA}}$ is determined to be 39% by equation (38). This value is in agreement with the experimental data of Vickers hardness measurements on β-Si$_3$N$_4$ single crystals either grown from Si melt or present in sintered Si$_3$N$_4$ ceramics. The dissolution of Al and O leads to a decrease in the hardness of the β-Si$_3$N$_4$ lattice by the formation of β-Sialon (Si$_6$−$z$Al$_z$O$_z$N$_8$−$z$) [181]. Tanaka et al [182] found that the decrease in hardness with the $z$ value can be expressed as

$$\text{HV} = 15.4 - 0.75z \text{ (GPa)}.$$

(39)

Therefore, the c-axis alignment of β-Si$_3$N$_4$ grains combined with the use of a small amount of an Al-free sintering...
additive is expected to result in the formation of high-hardness β-Si₃N₄ ceramics in the plane parallel to the grain alignment.

6.2. Elastic modulus

The elastic modulus is an important physical parameter for understanding the mechanical properties and structural applications of Si₃N₄ ceramics. The Young’s moduli of Si₃N₄ ceramics were reported to be in the range of 300–330 GPa, depending on porosity, grain boundary phases, texture and the relative contents of α and β-phases [180]. Lee and Bowman [110] first reported the elastic modulus anisotropy in hot-pressed Si₃N₄ ceramics, which showed Young’s moduli of 332 and 315 GPa parallel and perpendicular to the HP direction, respectively. The authors supposed that the small anisotropy in the Young’s modulus is due to the low anisotropy in the Young’s modulus of the β-Si₃N₄ single crystal. This assumption has been confirmed by Hay et al [180] using nanoindentation measurements on individual large elongated hexagonal β-Si₃N₄ grains, as shown in figure 25. In their work, the measured elastic moduli are specifically defined as ‘indentation moduli’ rather than ‘Young’s moduli’ because the data analysis procedures used to obtain indentation moduli from the nanoindentation load-displacement data are based on contact mechanics solutions for elastically isotropic media and therefore only formally apply to isotropic materials. For anisotropic materials, the indentation modulus is generally different from the Young’s modulus, and its value depends on all the elastic constants of the materials in a complex way, as shown in figure 25. In the stiffest direction (α = 0°, i.e. the c-axis of the crystal) the indentation modulus underestimates the Young’s modulus of the single crystal by about 20%, and in the most compliant direction it overestimates Young’s modulus by about 10%.

Figure 25 provides a measure of the magnitude of elastic anisotropy in β-Si₃N₄. The estimated Young’s moduli for β-Si₃N₄ are 280 and 540 GPa in the directions perpendicular and parallel to the c-axis, respectively, indicating that the elastic anisotropy of an elongated hexagonal β-Si₃N₄ crystal is highly intrinsic. If the intrinsic elastic anisotropy factor for a β-Si₃N₄ crystal, \( f_{\text{IEA}} \), is defined as

\[
f_{\text{IEA}} = \left( 1 - \frac{E_{\perp}}{E_{\parallel}} \right) \times 100\%,
\]

where \( E_{\perp} \) and \( E_{\parallel} \) are the Young’s moduli in the directions perpendicular and parallel to the c-axis of the β-Si₃N₄ crystal, respectively. Thus, \( f_{\text{IEA}} \) is determined to be 48%. The elastic anisotropy was also reported by Kondo et al [183] in textured porous Si₃N₄ ceramics prepared by partial SF, in which the Young’s modulus is lower in the direction perpendicular to the c-axis of elongated grains than in the direction parallel to the c-axis. Therefore, the c-axis alignment of elongated β-Si₃N₄ grains should reduce the Young’s modulus in the direction perpendicular to the grain alignment, thereby giving Si₃N₄ ceramics excellent strain tolerance [184]. In addition to the intrinsic elastic anisotropy, the β-Si₃N₄ grain morphology, such as the aspect ratio, should also affect the elastic anisotropy of textured Si₃N₄ ceramics [185]. To obtain a full understanding of the elastic anisotropy behavior, further studies are needed, including experimental and model simulation studies.

6.3. Strength and toughness

6.3.1. RT strength and toughness. It has been well documented that texturing leads to anisotropy in the strength and toughness of Si₃N₄ ceramics. This offers a unique
opportunity to maximize the strength and toughness in the direction perpendicular to the grain alignment. Figure 26 shows the important progress achieved in improving the strength and toughness of untextured and textured Si$_3$N$_4$ ceramics. Compared with untextured Si$_3$N$_4$, textured Si$_3$N$_4$ has the capability of reaching a bending strength of up to 1.4 GPa and a fracture toughness of up to 14 MPa m$^{1/2}$ in the direction perpendicular to the grain alignment. The grain orientation provides more opportunities for crack deflection and bridging in the crack propagation plane perpendicular to the grain alignment, thereby higher fracture toughness [15, 41–43, 186]. On the one hand, the high bending strength is due to the more uniform 2-D distribution of larger elongated grains, thereby preventing the formation of large defects such as clusters. On the other hand, the high fracture toughness in the same direction also contributes to the high strength, as indicated by the relationship between bending strength ($\sigma_b$) and fracture toughness ($K_{IC}$) [187]

$$\sigma_b = \frac{K_{IC}}{\sqrt{\pi a}},$$

(41)

where $a$ is half the crack length. The higher toughness allows textured Si$_3$N$_4$ to have higher tolerance to flaws generated during machining, thereby improving the mechanical reliability. Hirao et al [43] developed c-axis aligned Si$_3$N$_4$ that exhibited a Weibull modulus of 46. Moreover, Ohji et al [188] observed a steep R-curve behavior for textured Si$_3$N$_4$, as shown in figure 27, which is believed to be advantageous for avoiding catastrophic fractures and for increasing the Weibull modulus by narrowing the strength distribution. Pezzotti et al [189] used Raman microprobe spectroscopy to reveal that crack-face bridging is the most effective mechanism for toughening Si$_3$N$_4$ ceramics. The macroscopic fracture behavior of untextured and textured Si$_3$N$_4$ can be modeled as a function of the microscopic maps of bridging tractions; the higher the bridging stress, the steeper the R-curve.

The bending strength anisotropy increases with increased degree of orientation. The bending strength anisotropy factor, $f_{BSA}$, is defined as

$$f_{BSA} = \left( 1 - \frac{\sigma_{b\perp}}{\sigma_{b\parallel}} \right) \times 100\%,$$

(42)

where $\sigma_{b\perp}$ and $\sigma_{b\parallel}$ are the bending strengths in the directions perpendicular and parallel to the grain alignment, respectively. Figure 28 shows the bending strength anisotropy versus the bending strength $\sigma_{b\perp}$ of c-axis aligned Si$_3$N$_4$. The bending strength anisotropy increases with increased strength $\sigma_{b\perp}$.

The Vickers indentation fracture (VIF) method has been widely used for determining the fracture toughness of untextured and textured Si$_3$N$_4$ ceramics. For textured Si$_3$N$_4$, the toughness anisotropy can be determined by the anisotropic crack propagation between the directions perpendicular and parallel to the grain alignment, as typically shown in figure 29. In the direction perpendicular to the grain alignment ($\perp$ c-axis), the cracks are shorter and toughness is higher because more energy is released per unit crack length, compared with the direction parallel to the grain alignment ($\parallel$ c-axis). This means that in the VIF method, one indentation point can produce one toughness anisotropy value. In contrast, the single-edge-precracked-beam (SEPB) and the single-edge-V-notched-beam (SEVNB) methods require two samples to produce one toughness anisotropy value, because only one precracked plane is operated on for each sample. The fracture toughness anisotropy factor, $f_{FTA}$,
Table 12. Fracture toughness anisotropy in textured Si₃N₄ as reported in the literature.

| Authors          | Processing method | Key factor | DOT<sup>a</sup> | Method<sup>b</sup> | K<sub>IC</sub> (MPa m<sup>1/2</sup>)<sup>c</sup> | f<sub>FTA</sub> |
|------------------|-------------------|------------|------------------|-------------------|---------------------------------|--------------|
| Lee and Bowman   | HW                | HP         |                 | VIF               | 5.72                            | 19           |
|                  |                   | HF_SP      | ↓                |                   | 8.41                            | 30           |
| Xie et al [64]   | HF-PSC            | True strain: |                 |                   | 3.32                            | 19           |
|                  |                   | 0.3        | VIF             |                   | 3.22                            | 19           |
|                  |                   | 0.5        |                   |                   | 2.67                            | 31           |
|                  |                   | 0.7        | ↓                |                   | 2.51                            | 46           |
|                  |                   | 0.95       |                  |                   | 2.54                            | 52           |
| Santos et al [126]| HP              | Additive (vol%) |                 | VIF               | 6.6                             | 21           |
|                  |                   | 5          |                  |                   | 5.2                             | 14           |
|                  |                   | 10         | ↑                |                   | 5.6                             | 14           |
|                  |                   | 15         |                  |                   | 5.8                             | 9            |
|                  |                   | 20         |                  |                   | 5.9                             | 5            |
| Imamura et al [56]| TCA            | Seed type: |                 | SEPB              | 10.4                            | 32           |
|                  |                   | w = 0.44 µm, R = 3.1 | | [] | 7.1                             | []           |
|                  |                   | w = 1.29 µm, R = 4 | | [] | 7.3                             | 37           |
|                  |                   | w = 0.47 µm, R = 4.2 | | [] | 7.2                             | 42           |
| Muscat et al [142]| EA            | β-Si₃N₄ whisker seed | | VIF               | 13.5                            | 66           |
| Nakamura et al [63]| EA            | β-Si₃N₄ rod seed | | VIF               | 9.7                             | 60           |
| Park et al [57]  | TCA              | β-Si₃N₄ whisker seed | | VIF               | 8.52                            | 40           |

<sup>a</sup> DOT = degree of texture, ↑↓ means increased DOT upon varying parameters from up to down and vice versa.

<sup>b</sup> VIF = Vickers indentation fracture, SEPB = single-edge-precracked-beam method.

<sup>c</sup> ⊥ and || denote the directions perpendicular and parallel to the alignment of elongated grains, respectively.

Figure 28. Bending strength anisotropy versus bending strength (σ₅₄) in the direction perpendicular to the grain alignment, corresponding to the direction of the tensile stress plane parallel to the grain alignment.

is defined as

\[
f_{\text{FTA}} = \left(1 - \frac{K_{\text{IC}||}}{K_{\text{IC}\perp}}\right) \times 100\%,
\]

where \(K_{\text{IC}||}\) and \(K_{\text{IC}\perp}\) are the fracture toughnesses perpendicular and parallel to the grain alignment, respectively.

6.3.2. HT strength. It is well-known that the HT strength of Si₃N₄ ceramics is mainly determined by the softening temperature of the grain boundary phase. Therefore, two common methods are used for improving HT strength: choosing a more refractory additive composition and crystallizing grain boundary amorphous phases by a postheating treatment. Table 13 lists the important progress in a higher \(f_{\text{FTA}}\) than the \(a, b\)-axis aligned texture. Because of the limited \(a, b\)-axis orientation, HP produces the lowest \(f_{\text{FTA}}\), whereas because of the high \(c\)-axis orientation, EA produces the highest \(f_{\text{FTA}}\); the TCA and HF methods produce intermediate values of \(f_{\text{FTA}}\). Therefore, EA is the most efficient method for developing textured Si₃N₄ with the highest anisotropic mechanical properties.
Table 13. HT strength of untextured and textured Si$_3$N$_4$ as reported in the literature.

| Authors                      | Additive Composition                     | Microstructure       | Atmosphere  | RT  | 1400 | 1500 | $\%_{\text{ retention}}$ |
|------------------------------|------------------------------------------|----------------------|-------------|-----|------|------|---------------------------|
| Lu and Huang [30]           | Y$_2$O$_3$–Al$_2$O$_3$                  | Randomly aligned     | N$_2$       | 85  | 375  | 350  | 44                        |
| Park et al. [68]            | Y$_2$O$_3$–Al$_2$O$_3$                  | Randomly aligned     | N$_2$       | 83  | 203  | 180  | 404                       |
| Nahidim et al. [24]         | Y$_2$O$_3$–SiO$_2$                      | Highly c-axis aligned| N$_2$       | 870 | 977  | 100  | 647                       |
| Park et al. [25]            | Y$_2$O$_3$–SiO$_2$                      | Weakly c-axis aligned| N$_2$       | 90  | 50   | 87   | 70                        |
| Guo et al. [31]             | Lu$_2$O$_3$–SiO$_2$                     | Weakly $\beta$ aligned| N$_2$       | 95  | 44   | 70   | 50                        |
| Kondo et al. [129]          | Lu$_2$O$_3$–SiO$_2$                     | Highly $\beta$ aligned| Air         | 90  | 60   | 696  | 643                       |
| Zeng et al. [72]            | Lu$_2$O$_3$–SiO$_2$                     | $\sim$ $\beta$-Si$_3$N$_4$ | Air         | 930 | 982  | 940  | 556                       |

Note:
- Four-point bending test.
- Three-point bending test.
- $\%_{\text{ retention}}$ is the percentage of retention with respect to the RT strength.

6.4. Fracture energy

Fracture energy is the most significant factor determining the strength of Si$_3$N$_4$ ceramics. The relationship between fracture energy ($\gamma$) and strength ($\sigma$) is expressed by Griffith’s fracture equation [187]

$$\sigma = A \sqrt{\frac{\gamma E}{C}},$$

where $E$ is the Young’s modulus, $C$ is the length of the crack that initiates the fracture and $A$ is a dimensionless number that depends on the mode of stressing, specimen configuration and dimensions, and the type of crack under consideration. The fracture energy can be understood as being the rate at which energy is absorbed by the growth of the crack. Equation (44) indicates that a higher fracture energy can result in a material having a higher strength. Si$_3$N$_4$ ceramics are well-known for high strength in comparison with other structural ceramics such as Al$_2$O$_3$, AlN and SiC. The high strength is closely related to the higher fracture energy, because the interlocking microstructure of elongated $\beta$-Si$_3$N$_4$ grains increases the
Table 14. Fracture energy of untextured and textured Si$_3$N$_4$ as reported in the literature.

| Fracture energy (J m$^{-2}$) | HT ($^\circ$C) | RT | Authors | Additive | Microstructure | Atmosphere |
|-----------------------------|---------------|-----|---------|-----------|----------------|------------|
| 1300                        | 62            | 52  | Lange [8] | MgO       | Weakly $a$, $b$-axis aligned | Air        |
| 1500                        | 75            | 72  | Kondo et al [61] | Y$_2$O$_3$−Al$_2$O$_3$ | Randomly aligned | Air        |
| 720                         | 80            | 720 | Kondo et al [128] | Yb$_2$O$_3$−SiO$_2$ | Weakly $a$, $b$-axis aligned | Air        |
| 753                         | 48            | 570 | Zeng et al [70] | Lu$_2$O$_3$−SiO$_2$ | Randomly aligned | Air        |
| 727                         | 118           | 753 | Zeng et al [70] | Lu$_2$O$_3$−SiO$_2$ | Highly $a$, $b$-axis aligned | Air        |

a Double-cantilever (DC) method with a crosshead speed of 0.5 mm min$^{-1}$.

b Chevron-notched-beam (CNB) method with a displacement rate of 0.005 mm min$^{-1}$.

c CNB method with a displacement rate of 0.01 mm min$^{-1}$.

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6.5. Creep behavior

The creep behavior of Si$_3$N$_4$ ceramics has been extensively studied, and a review article of the processes involved in this area has been published recently [190]. Generally, the creep deformation mechanisms of Si$_3$N$_4$ ceramics are attributed to viscous flow, solution precipitation, cavitation and shear thickening, depending on the various factors involved in determining the characteristics of Si$_3$N$_4$ ceramics and the conditions of creep tests [190]. However, studies concerning the role of texture in creep deformation behavior are very limited. Yoon et al [48] studied the effect of the grain orientation on the compressive creep deformation of hot-pressed Si$_3$N$_4$ in air at temperatures of 1300–1400 $^\circ$C and pressures of 30–100 MPa in air. They observed that the creep rate in the loading direction parallel to the HP was higher than that in the loading direction perpendicular to the HP. The stress exponent of the creep rate was determined to be about 1 and the apparent activation energy in both parallel and perpendicular directions at the temperature range of 1300–1400 $^\circ$C is about 500 kJ mol$^{-1}$. This suggests that the creep deformation is mainly determined by diffusion-controlled solution precipitation. Combined energy dissipation by the frictional sliding of one surface against another during crack propagation.

Lange [8] revealed that the fracture energy of Si$_3$N$_4$ ceramics could be increased not only by tailoring the grain morphology, but also by developing the texture or grain orientation. He first reported that the fracture energy is about 16% higher in the direction perpendicular to the $c$-axis of elongated $\beta$-Si$_3$N$_4$ than in the direction parallel to the $c$-axis in hot-pressed Si$_3$N$_4$. Table 14 lists the fracture energy of untextured and textured Si$_3$N$_4$ ceramics as reported in the literature. At RT, the texture design is more efficient for increasing the fracture energy than the additive composition. The unidirectional aligned texture results in a higher fracture energy than the $a$, $b$-axis aligned texture in the direction perpendicular to the $c$-axis of the elongated grains at both RT and HTs, corresponding to the higher anisotropy. The testing atmosphere has no significant effect on the fracture energy at HTs. However, the anisotropy of fracture energy becomes small at HTs. The fracture energy anisotropy factor, $f_{FEA}$, is defined as

$$f_{FEA} = \left(1 - \frac{\gamma_\parallel}{\gamma_\perp}\right) \times 100\%,$$

where $\gamma_\perp$ and $\gamma_\parallel$ are the fracture energies in perpendicular and parallel to the $c$-axis of elongated $\beta$-Si$_3$N$_4$ grain, respectively. Thus, $f_{FEA}$ decreases significantly from 61% at RT to 7% at 1500 $^\circ$C, according to the data of Zeng et al [70]. The large increase in fracture energy by grain orientation is attributed to the effects of grain bridging and grain pullout. The fracture energy anisotropy is responsible for the anisotropy of the RT strength of textured Si$_3$N$_4$. At HTs, the higher fracture energy is due to the softening of the secondary glassy phases and the melting of secondary crystalline phases, resulting in enhanced grain pullout.
with TEM observation, the authors attributed the observed anisotropic creep behavior to the difference in the rate of the solution-precipitation step. Compared with the loading direction parallel and perpendicular to the HP, the direction parallel to the HP exhibits a higher solution-precipitation rate because of the unstrained basal (002) plane. Santos et al [71] also reported similar anisotropic creep behavior in hot-pressed Si$_3$N$_4$ between the directions parallel and perpendicular to the HP.

Kondo et al [60] studied the tensile creep behavior of untextured and textured Si$_3$N$_4$ ceramics prepared by GPS and SF, respectively, at temperatures of 1200–1250 °C in air. They found that the creep rate along the grain alignment direction in the textured sample (c-axis) was lower about one order of magnitude than that of the untextured sample. The stress exponents of the untextured and textured samples at 1200 °C were 2.1 and 2.6, respectively, and the value for the textured sample at 1250 °C was 3.6. This reveals that the grain alignment can improve the tensile creep resistance of Si$_3$N$_4$ ceramics in the grain alignment direction. The authors proposed several possible mechanisms for achieving significantly improved tensile creep resistance by the c-axis alignment of elongated grains: (i) increasing the sliding resistance to the grain boundaries, (ii) suppressing the formation of cavities and (iii) inhibiting viscous flow by providing a large number of film-free boundaries between well-aligned elongated grains. Hirao [75] also reported that c-axis aligned Si$_3$N$_4$ exhibits higher flexural creep resistance in the direction perpendicular to the grain alignment in comparison with untextured samples. Furthermore, the creep resistance increases with increasing amount of aligned elongated grains, i.e., degree of orientation, as shown in figure 30. The textured material can possess flexural creep resistance that is 1–2 magnitudes higher than that of the untextured material. The results suggest that texture design is an effective method for improving the creep resistance of Si$_3$N$_4$ ceramics. To produce highly creep-resistant Si$_3$N$_4$, an effective processing strategy should involve a combination of texture design, effective additive composition and optimized processing parameters.

6.6. Tribological and wear resistance

Miller and Bowman [46] first observed through sliding-ball tests that textured Si$_3$N$_4$ exhibited less damage in the plane perpendicular to the grain alignment than in the direction parallel to the grain alignment and that cracks tend to propagate in the sliding direction. Liang et al [54, 55] investigated the effects of the degree of orientation and sintering temperature on the friction and wear behavior of textured Si$_3$N$_4$ ceramics using ball-on-disk reciprocating sliding tests with Si$_3$N$_4$ balls under dry friction. They showed that a highly c-axis aligned sample exhibited much lower wear resistance in the direction parallel to the grain alignment than in the direction perpendicular to the grain alignment, and the higher degree of orientation resulted in lower wear resistance parallel to the sliding direction but higher wear resistance perpendicular to the sliding direction. An increased sintering temperature tends to degrade the wear resistance both parallel and perpendicular to the grain alignment. Furthermore, they also showed that the cross-alignment design of elongated grains allowed the wear resistance of this material to be between those of the highly c-axis aligned material in the directions parallel and perpendicular to the grain alignment. The effect of grain orientation on the friction coefficient depends on the load, but the dependence is not sufficiently strong to affect the wear resistance.

Nakamura et al [63, 66] conducted comparative studies of tribological and wear behavior between textured and untextured Si$_3$N$_4$ under dry friction using block-on-ring tests. They observed that for both Si$_3$N$_4$ and stainless steel rings, the unidirectionally aligned Si$_3$N$_4$ exhibited higher wear resistance in the plane perpendicular to the elongated grain alignment than in the direction both parallel and perpendicular to the grain alignment, and also higher wear resistance than the untextured sample for all sliding conditions. In the case of the Si$_3$N$_4$ ring, the wear resistance is greater perpendicular to the grain alignment than parallel to the grain alignment [63], whereas in the case of the stainless-steel ring, the wear resistance depends on the sliding conditions [66]. Table 15 lists the data of the friction coefficient and wear rate of unidirectionally aligned and randomly aligned Si$_3$N$_4$ ceramics for a Si$_3$N$_4$ ring, as reported by the authors.

The wear behavior of ceramics is dominated by the microfracture mechanism [191]. The sliding process generates compressive and tensile stresses near the contact surface, resulting in the debonding of the elongated β-Si$_3$N$_4$ grains from the small matrix by the initiation and propagation of microcracks. Thus, different debonding strengths lead to different wear behaviors. For highly c-axis aligned Si$_3$N$_4$, during the sliding process a lower number of elongated grains are pulled out in the direction perpendicular to the grain alignment than in the direction parallel to the grain alignment because of the higher crack propagation resistance, thereby
resulting in a lower wear rate. Comparing the directions parallel and perpendicular to the grain alignment, the plane perpendicular to the grain alignment exhibits the lowest probability of grain pullout; therefore, it has the lowest wear rate. The anisotropic friction and wear mechanisms in textured Si$_3$N$_4$ are schematically illustrated in figure 31. The worn surface in the sliding direction parallel to the grain alignment (figure 31(a)) is typically covered by a thick layer of wear debris, and the worn surface in the sliding direction perpendicular to the grain alignment (figure 31(b)) is covered by a small amount of wear debris, whereas the worn surface in the plane perpendicular to the grain alignment (figure 31(c)) is reasonably smooth.

6.7. Erosion resistance

Lim et al [67] studied the erosion behavior of unidirectionally aligned Si$_3$N$_4$ by a gas-blast-type erosion tester using SiC grit at impact angles from 45 to 90°, as schematically shown in figure 32. They found that the direction parallel to the grain alignment exhibits higher erosion resistance (or lower erosion rate) than the direction perpendicular to the grain alignment for all tested impact angles. They attributed this erosion anisotropy to the higher probability of grain pull-out because of the ease of lateral crack formation in the direction perpendicular to the elongated grain alignment, thereby resulting in the greater amount of material removal.

Zhang et al [69] conducted a comparative study of the erosion behavior of textured and untextured Si$_3$N$_4$ by the same testing technique using SiC grit with two impact angles of 30 and 90°. They found that the textured Si$_3$N$_4$ exhibits lower erosion resistance than the untextured Si$_3$N$_4$ at both impact angles, as listed in table 16. Generally, the erosion rate of ceramics should have a strong inverse dependence on the fracture toughness but a much weaker dependence on the hardness. According to theoretical predictions, highly unidirectionally aligned Si$_3$N$_4$ should have exhibited better erosion resistance in the direction perpendicular to the grain alignment than the randomly aligned Si$_3$N$_4$ because of the higher fracture resistance. The authors attributed this to the microstructure effect. The unidirectionally aligned microstructure promotes grain pullout, whereas the randomly aligned microstructure provides an interlocking effect that hinders it. In addition, because of the higher incidence of grain pullout, the direction parallel to the grain alignment exhibited a higher erosion rate than the direction perpendicular to the grain alignment, which is contrary to the findings of Lim et al [67]. This contradiction concerning the effect of grain
Figure 32. Schematic illustrations of impact direction of erosive SiC particles with respect to the alignment of elongated grains: (a) impact perpendicular to the grain alignment and (b) impact parallel to the grain alignment. $\alpha$ is the impact angle.

Table 16. Erosion data of UASN and RASN as reported by Zhang et al [69].

| Material | Impact angle (°) | Test direction | Erosion rate $(m^3 kg^{-1}) \times 10^{-3}$ |
|----------|------------------|----------------|------------------------------------------|
| UASN     | 30               | $\parallel c$-axis | 26                                       |
|          | 90               | $\perp c$-axis    | 24                                       |
| RASN     | 30               | $\parallel c$-axis | 13                                       |
|          | 90               | $\parallel c$-axis | 37                                       |

orientation on erosion resistance may be due to the testing conditions. To obtain a full understanding of the effect of texture on the erosion behavior, more erosion conditions should be considered, such as the characteristics of erosive particles, the velocity and dosage of erosive particles and the distance between the nozzle and sample.

6.8. Hertzian contact damage

The Hertzian indentation technique has been used to characterize the contact damage of Si$_3$N$_4$ ceramics, which is crucial for some structure applications, e.g., frictional or erosive environments where the surfaces undergo impact due to a load. Lee et al [192, 193] studied the effect of the microstructure on the Hertzian contact damage of Si$_3$N$_4$ ceramics. They showed that there is competition between brittle (ring or cone fracture) and quasi-plastic damage (distributed shear-activated microfaulting) modes. The microstructure with relatively equiaxed grains exhibits damage in the form of classical cone cracks, whereas the microstructure with large elongated grains exhibits damage characterized by grain-localized microfailures within a subsurface, indicating higher contact damage resistance. Furthermore, in several studies it has been revealed that the grain orientation has a strong effect on the contact damage behavior of Si$_3$N$_4$ ceramics [46, 144]. Figure 33 shows typical Hertzian contact damage patterns of weakly $a$, $b$-axis and highly $c$-axis aligned Si$_3$N$_4$ ceramics, which show the random and unidirectional alignment of the $c$-axis in testing plane A, respectively [144]. Compared with the $a$, $b$-axis aligned sample, the $c$-axis aligned sample exhibits fewer and shallower surface ring cracks, which have an elliptical shape. This elliptical shape is due to the deflection of crack trajectories perpendicular to the interface caused by unidirectionally aligned elongated grains quasi-parallel to the bonding interface (plane B). It is clear that beneath the contact area, the unidirectional alignment of elongated grains causes the contact damage to change from cone cracks to a combination of cone cracks and a subsurface damage zone. This subsurface damage zone increases in size with the indentation load. Moreover, in the subsurface zone, multiple twins/slips are observed at the large Si$_3$N$_4$ seeds, together with debonding at seed interfaces. The twins are found to be oriented at about 45° to the contact axis, which is the direction of maximum shear stress [46]. The results indicate that the grain alignment affects the contact damage behavior in two ways: (i) conferring an elliptical shape to the radial surface cracks and (ii) promoting the multiple twin/slip formation at the large elongated grains within the high-shear-strain region.

6.9. Thermal conductivity

In 1995, Haggerty and Lightfoot [194] first pointed out that Si$_3$N$_4$ has an intrinsic thermal conductivity of over 200 W m$^{-1}$ K$^{-1}$ at RT. Si$_3$N$_4$ has since been listed as a new high-thermal-conductivity ceramic material, similar to AlN and SiC. A large amount of experimental work has resulted in the establishment of a state-of-the-art method of producing $\beta$-Si$_3$N$_4$ ceramics with thermal conductivities of over 100 W m$^{-1}$ K$^{-1}$, including the utilization of effective Al-free sintering additives, HT (and/or long-time) sintering/annealing and grain orientation techniques. The improvement in thermal conductivity is primarily attributed to the purification of $\beta$-Si$_3$N$_4$ grains, such as by removing the dissolved oxygen in the $\beta$-Si$_3$N$_4$ lattice, because lattice defects cause the scattering of phonons that dominate heat conduction. There are several review articles dealing with the progress achieved in this field [195–197]. Because of the combination of
high thermal conductivity, superior mechanical properties and an excellent insulating property, $\beta$-$\text{Si}_3\text{N}_4$ has been regarded as a promising substrate material for semiconductor devices. Enhanced thermal conductivity is also helpful for expanding its application to structural materials such as engine components and heat exchangers, where high thermal-shock resistance is required to improve the long-term reliability.

Textured $\beta$-$\text{Si}_3\text{N}_4$ ceramics exhibit an interesting thermal anisotropy: the thermal conductivity is higher in the direction of the elongated grain alignment than in the direction perpendicular to the grain alignment $[39, 45, 47, 52, 53, 58, 59, 62, 65, 72]$. Thus, texturing has become another important method for improving the thermal conductivity of $\beta$-$\text{Si}_3\text{N}_4$ ceramics. Below we will discuss the thermal anisotropy of textured $\beta$-$\text{Si}_3\text{N}_4$ both theoretically and experimentally.

6.9.1. Theoretical analysis. To analyze the thermal anisotropy of $\beta$-$\text{Si}_3\text{N}_4$ ceramics, it is necessary to understand the thermal anisotropy of a single $\beta$-$\text{Si}_3\text{N}_4$ crystal. Several studies have been carried out to predict the intrinsic thermal conductivity of a single $\beta$-$\text{Si}_3\text{N}_4$ crystal $[194, 196, 198, 199]$, as summarized in table 17. However, the thermal anisotropy was only taken into account in the work of Hirosaki $et\ al$ $[198]$. Using the molecular dynamics method, they predicted that a single $\beta$-$\text{Si}_3\text{N}_4$ crystal has intrinsic thermal conductivities of 170 and 450 W m$^{-1}$ K$^{-1}$ along the $a$- and $c$-axis directions, respectively. To describe the thermal anisotropy of a single $\beta$-$\text{Si}_3\text{N}_4$ crystal, the factor, $f_{\text{ITA}}$, is defined as

$$f_{\text{ITA}} = \left(1 - \frac{k_a}{k_c}\right) \times 100\%,$$  \hspace{1cm} (46)
where $\kappa_c$ and $\kappa_a$ are the thermal conductivities along the c- and a-axis directions, respectively. Thus, for an ideal single crystal, $f_{ITA}$ is 62%. Moreover, such strong thermal anisotropy has also been demonstrated experimentally in actual rodlike $\beta$-Si$_3$N$_4$ crystals to be independent of the diameter by Li et al [200], as shown in table 17. The fact that the theoretical value agrees well with the experimental value suggests that the thermal anisotropy of $\beta$-Si$_3$N$_4$ crystals is highly intrinsic.

Kitayama et al [53] developed a modified Wiener model to describe the thermal conductivity of $\beta$-Si$_3$N$_4$ ceramics that contain oriented rodlike grains surrounded by a continuous grain boundary phase. They simply treated the textured microstructure as an idealized 2D microstructure composed of perfectly oriented $\beta$-Si$_3$N$_4$ grains along the c-axis direction. Using this new model, they predicted thermal conductivities as a function of grain boundary film thickness and grain size, which were in good agreement with experimental results. At the same time, they also showed that the thermal anisotropy is affected by the grain boundary film thickness and grain size. However, in their work, the intrinsic thermal anisotropy of $\beta$-Si$_3$N$_4$ grains was not considered. In this article, we use the Kitayama model to analyze the thermal anisotropy of textured $\beta$-Si$_3$N$_4$, as expressed by the following equations

$$\frac{1}{\kappa_i} = \frac{l}{l + \delta} \times \frac{w + \delta}{w + \frac{w}{\delta} (2\kappa_c + 2\kappa_g + 2f_p (\kappa_g - \kappa_c)) + \delta \kappa_g} + \frac{\delta}{l + \delta \kappa_g},$$

$$\frac{1}{\kappa_\perp} = \frac{w}{w + \delta} \times \frac{l + \delta}{l + \delta \kappa_g} \times \frac{\kappa_g (2\kappa_c + 2\kappa_g + 2f_p (\kappa_g - \kappa_c)) + \delta \kappa_g}{\kappa_g (2\kappa_c + 2\kappa_g + 2f_p (\kappa_g - \kappa_c)) + \delta \kappa_g} + \frac{\delta}{w + \delta \kappa_g},$$

where $\kappa_i$ and $\kappa_\perp$ are defined as the bulk thermal conductivities parallel and perpendicular to the c-axis (long-axis) of $\beta$-Si$_3$N$_4$ grains, respectively; $\kappa_c$ and $\kappa_g$ are the thermal conductivities of the $\beta$-Si$_3$N$_4$ crystal and the grain boundary glassy phase, respectively; $w$ and $l$ are the width and length of elongated $\beta$-Si$_3$N$_4$ grains, respectively; $\delta$ is the grain boundary film thickness; and $f_p$ is the volume fraction of the glass pockets, given by

$$f_p = x = \frac{\delta (w + l)}{w l} (1 - x),$$

where $x$ is the volume fraction of the glassy phase. To describe the thermal anisotropy of textured $\beta$-Si$_3$N$_4$ ceramics, the factor, $f_{ITA}$, is defined as

$$f_{ITA} = \left( 1 - \frac{\kappa_\perp}{\kappa_i} \right) \times 100\%.$$ 

A larger value of $f_{ITA}$ implies stronger thermal anisotropy.

First, to study the effect of only the aspect ratio of $\beta$-Si$_3$N$_4$ grains on the thermal anisotropy of c-axis oriented $\beta$-Si$_3$N$_4$ ceramics, only the isotropic thermal conductivity is considered for the $\beta$-Si$_3$N$_4$ single crystal. The values used in the calculations are $w = 1 \mu m$, $\kappa_c = 180 W m^{-1} K^{-1}$ [45, 201], $\kappa_g = 1 W m^{-1} K^{-1}$ [202] and $\delta = 1 nm$. Figure 34(a) shows that the elongated shape leads to thermal anisotropy. The thermal anisotropy initially increases rapidly with increasing aspect ratio, then remains almost constant above a critical value ($\approx 4$). Obviously, the isotropic grain shape ($R = 1$) does not lead to thermal anisotropy. In addition, the thermal anisotropy decreases slightly with increasing volume fraction of the glassy phase.

Second, to study the effect of the degree of orientation on the thermal anisotropy, we use the varied intrinsic thermal anisotropy in the range of $f_{ITA} = 0$–62%. When $f_{ITA} = 0$, no orientation occurs; when $f_{ITA} = 62$%, the orientation is perfect. The values used in the calculations are $w = 1 \mu m$, $\kappa_\parallel = 180 W m^{-1} K^{-1}$ and $\kappa_\perp = \kappa_\parallel (1 - f_{ITA}/100) W m^{-1} K^{-1}$. Figure 34(b) shows the effect of the
6.9.2. Experimental results. In 1978, thermal anisotropy in β-Si₃N₄ ceramics was first reported for hot-pressed Si₃N₄ by Hirai et al [39]. They showed that the thermal conductivity along the direction parallel to the HP is 36% lower than that along the direction perpendicular to the HP at 20°C. Moreover, this value was found to decrease to 15% at 1300°C. Kitayama et al conducted a series of studies on the thermal conductivity of hot-pressed Si₃N₄, including the roles of the starting powders [53], lattice oxygen [58] and RE oxide additives [62], as summarized in figure 35. Figure 35(d) shows the two cases for measuring thermal conductivity by the laser flash method. Figure 35(a) shows that fine SN-E10 powder leads to a smaller grain size but a larger thermal anisotropy than coarse SN-E05 powder, primarily as a result of the larger aspect ratio, which is believed to increase the degree of orientation during HP [53]. However, the annealing time has little effect on the thermal anisotropy, indicating that the grain growth plays a less important role in the degree of orientation during annealing. The textured β-Si₃N₄, which was hot-pressed from β-Si₃N₄ powder in a shorter time without annealing, exhibits a substantially smaller grain size than the sample hot-pressed from SN-E05 powder followed by annealing, but both samples exhibit almost the same thermal anisotropy. Figure 35(b) shows the effect of the ionic radius of RE cations (RE³⁺) on the thermal anisotropy. It was found that during HP, the La₂O₃-doped sample exhibited the lowest thermal anisotropy, corresponding to the largest ionic radius of RE³⁺. This is attributed to the incomplete α→β phase transformation, thereby resulting in the lowest degree of orientation. However, prolonged annealing allows samples to exhibit an increase in the thermal anisotropy with increased ionic radius, indicating that the grain growth contributes slightly to the texture development. Annealing
Table 18. Thermal conductivity (κ) of highly textured Si₃N₄ produced by the TCA technique.

| Year | Authors | Si₃N₄ powder | Si₃N₄ seed | Sintering/annealing conditions | κ (W m⁻¹ K⁻¹) |
|------|---------|--------------|------------|-------------------------------|---------------|
| 1996 | Hirao et al. [45] | α-Si₃N₄ | 5 vol% 5 wt% Y₂O₃ | HP (1800 °C, 2 h), flow N₂ | 121 ± 7.5 |
| 1996 | Hirao et al. [45] | β-Si₃N₄ | 50 vol% 75 wt% Y₂O₃+ | HP (1940 °C, 0.5 h), 1 MPH N₂/P₂1FC | 137 ± 9.7 |
| 1999 | Hirosaki et al. [47] | α-Si₃N₄ | 5 vol% 0.5 mol% Y₂O₃ | HP (1940 °C, 2 h), 1 MPa N₂ | 70 ± 3.0 |
| 1999 | Watari et al. [52] | α-Si₃N₄ | 5 vol% 5 wt% Y₂O₃ | HP (1800 °C, 2 h), flow N₂ | 155 ± 6.7 |
| 2000 | Lee et al. [59] | α-Si₃N₄ | 10 vol% 0.1–15 μm | GPS (2000 °C, 4 h), 3 MPa N₂ | 28 ± 6.4 |

The thermal conductivity increases with increased Y₂O₃ : SiO₂ molar ratio, which seems to correlate with the increased grain size. Actually, this should be attributed to the increased aspect ratio with increased Y₂O₃ : SiO₂ molar ratio, thereby resulting in a higher degree of orientation during HP and subsequent annealing [62, 135]. However, it is found that thermal anisotropy is limited to below 30% in hot-pressed Si₃N₄ regardless of the processing conditions. This is attributed to the a, b-axis aligned texture, in which the β-Si₃N₄ elongated grains are randomly oriented along the c-axis.

Table 18 shows a summary of the reported thermal conductivities of highly c-axis aligned β-Si₃N₄ ceramics produced by the TCA technique. The c-axis aligned β-Si₃N₄ exhibits thermal anisotropy values of ≥47%, which are substantially higher than those of a, b-axis aligned β-Si₃N₄. The highest thermal anisotropy reaches 66%, almost the same as the intrinsic thermal anisotropy of single β-Si₃N₄ grains, suggesting that a nearly perfect orientation is developed in such a material. Evidently, the thermal anisotropy is insensitive to the composition and amount of additives, but is sensitive to the degree of orientation, as theoretically predicted. It is interesting to see that by using the improved TCA method to increase the degree of orientation of seed particles, GPS also can produce textured β-Si₃N₄ that exhibits thermal anisotropy as strong as that produced by HP and hot-isostatic pressing (HIP). However, GPS is a relatively inexpensive process. As a result, to maximize the thermal anisotropy in β-Si₃N₄ ceramics, a highly c-axis aligned texture is of particular importance. To obtain higher thermal conductivity, the other processing conditions should be controlled at the same time, such as the use of effective Al-free additives and sintering techniques.

As discussed above, textured Si₃N₄ ceramics exhibit anisotropy in their thermal, physical and mechanical properties, such as the hardness, elastic modulus, bending strength, fracture toughness, fracture energy, creep resistance, tribological and wear resistance, erosion resistance, contact damage resistance and thermal conductivity. The anisotropic properties result either from the intrinsic anisotropy or from the grain orientation or from both, as summarized in table 19. On the basis of previous experimental evidence, figure 36 shows a scheme of the directions and planes giving the maximum properties in the c-axis oriented β-Si₃N₄ ceramics, the directions giving the lowest and highest elastic modulus are also shown, because the required elastic modulus depends on the structural applications. Although it was reported that, corresponding to the directions perpendicular and parallel to the grain orientation, c-axis oriented β-Si₃N₄ exhibits higher bend and tensile creep resistance, respectively, than untextured β-Si₃N₄, the effect of the grain orientation on the bend and tensile creep resistance of textured β-Si₃N₄ is still
Table 19. Anisotropic properties of textured Si₃N₄ and their origins.

| Anisotropic properties | Origins                                      |
|------------------------|----------------------------------------------|
| Hardness               | Intrinsic anisotropy and grain orientation   |
| Elastic modulus        | Intrinsic anisotropy and grain orientation   |
| Bend strength          | Grain orientation                            |
| Fracture toughness     | Grain orientation                            |
| Fracture energy        | Grain orientation                            |
| Creep resistance       | Grain orientation                            |
| Tribological and wear resistance | Grain orientation |
| Erosion resistance     | Grain orientation                            |
| Contact damage resistance | Grain orientation |
| Thermal conductivity   | Intrinsic anisotropy and grain orientation   |

Figure 36. Scheme of the directions and planes giving the maximum properties in the c-axis oriented β-Si₃N₄, which have been clearly demonstrated experimentally in the literature. Note that the directions giving the highest and lowest elastic modulus are shown, because the required elastic modulus depends on the structural applications.

Figure 37. Schematic microstructure of a textured porous Si₃N₄ ceramic with unidirectionally aligned elongated β-Si₃N₄ grains and ellipsoidal pores.

7. Textured porous Si₃N₄

Textured porous Si₃N₄ is a type of novel porous ceramic composed of oriented elongated grains and anisotropic pores. This unique porous microstructure gives this material excellent mechanical properties [183, 184, 203-206] and thermal-mechanical properties [207], including strength, fracture toughness, fracture energy, crack growth resistance and thermal shock resistance. Figure 37 schematically illustrates the microstructure of textured porous Si₃N₄ with unidirectionally oriented elongated grains and elongated pores developed by Shigegaki et al [184] by tape casting using a β-Si₃N₄ whisker raw powder. They showed that this unidirectionally textured porous Si₃N₄ with a porosity of 14.4% exhibits superior strain tolerance to untextured dense Si₃N₄ because of the lower Young’s modulus and the same high strength. Moreover, this material exhibited a fracture energy of 490 J m⁻², about 7 times that of untextured dense Si₃N₄ [203].

7.1. Processing

The methods of processing textured porous Si₃N₄ are similar to those used for producing textured dense Si₃N₄, such as HW and TGG. The only difference is that the processing of textured porous Si₃N₄ requires incomplete densification by tailoring the Si₃N₄ raw powder, additive composition and sintering conditions. Nevertheless, two major methods have been commonly used in previous studies: tape casting and SF. In tape casting, β-Si₃N₄ whiskers or rodlike seed crystals are used as raw powders to produce unidirectionally textured porous Si₃N₄, because they tend to retard densification during sintering because of the smaller specific surface area. Inagaki et al [204] showed that a similar porous texture develops when β-Si₃N₄ whiskers and rodlike seed crystals are used as raw materials. In partial SF, α-Si₃N₄ raw powders are used to preferentially develop elongated β-Si₃N₄ grains during sintering, thereby resulting in a porous texture due to grain rotation during forging. Nevertheless, it is difficult to produce unidirectionally textured porous Si₃N₄ by partial SF compared with by tape casting. From the industrial point of view, tape casting is an efficient processing method for producing high-quality textured porous Si₃N₄. β-Si₃N₄ whisker powders are no longer commercially available because of safety considerations. However, rodlike seed crystals are very expensive as a result of the limited manufacturing capacity. Therefore, alternative raw powders should be considered, particularly commercially available α-Si₃N₄ raw powder in conjunction with a seeding method,
as used for the processing of textured dense Si$_3$N$_4$. Actually, the addition of large rodlike $\beta$-Si$_3$N$_4$ seed crystals tends to retard the densification of fine $\alpha$-Si$_3$N$_4$ compacts [13, 44, 56]. Thus, the porous texture can be tailored by optimizing the morphology and amount of seed crystals, the additive composition and processing conditions. The use of fugitive fibers should be an efficient method for tailoring an elongated pore morphology and alignment; they can undoubtedly be easily aligned together with rodlike $\beta$-Si$_3$N$_4$ seeds during tape casting.

7.2. Properties

It has been shown that pores can provide more bridging sites by detaching the small matrices from the large elongated grains, thereby resulting in greater crack growth resistance than in untextured dense Si$_3$N$_4$, as shown in figure 38. This is also the major reason why textured porous Si$_3$N$_4$ possesses a high fracture energy than other types of Si$_3$N$_4$ ceramics, such as textured dense Si$_3$N$_4$ and untextured dense and porous Si$_3$N$_4$, as illustrated in figure 39. The fracture energy is affected by porosity and texture. The fracture energy is higher in the direction perpendicular to the grain alignment than in the direction parallel to the grain alignment. The effect of porosity on the fracture energy appears to depend on the texture. In the direction parallel to the grain alignment, the porosity always tends to reduce the fracture energy. However, in the direction perpendicular to the grain alignment, up to a porosity of $\sim$15%, the fracture energy increases with increasing porosity, but above this value the fracture energy decreases. This suggests that 15% may be the critical porosity for the fracture energy of textured porous Si$_3$N$_4$. However, no linear relationship in the semi-log plots of fracture energy against porosity can be found.

In contrast to the fracture energy, the reported data for the strength, toughness and Young’s modulus are found to have a linear relationship with porosity in semi-log plots, as illustrated in figures 40–42, respectively, i.e., the dependences of these mechanical properties on porosity can be described by the general formula [208]

$$P_{\phi} = P_{\phi 0} e^{-b \Phi}, \quad (51)$$

where $P_{\phi}$ and $P_{\phi 0}$ are the properties of the materials with and without pores, respectively, $b$ is the parameter determined by the pore characteristics and $\Phi$ is the volume fraction of the pores instead of the volume percentage. Table 20 shows a summary of the fitting results of bending strength, fracture toughness and Young’s modulus against porosity using equation (51). It is found that there is a similar value of $b = \sim$5 in the bending strength for the directions perpendicular and parallel to the grain alignment. Because the bending strength data in figure 40 was obtained from unidirectionally textured porous Si$_3$N$_4$ materials, it appears to suggest that the parameter $b$ is independent of the grain
Table 20: Fitting results of bending strength, fracture toughness and Young’s modulus by equation (51).

| Property                     | Equation          | $b^a$ | Property without porosity |
|------------------------------|-------------------|-------|---------------------------|
| Bending strength $\sigma_b$  | $\sigma_b = \sigma_{b0}e^{a+b}$ | 4.6   | $\sigma_{b0,\perp} = 1863$ MPa, $\sigma_{b0,\parallel} = 1096$ MPa |
| Fracture toughness $K_{IC}$  | $K_{IC} = K_{IC0}e^{a+b}$   | 2.8   | $K_{IC0,\perp} = 20$ MPa m$^{1/2}$ |
| Young’s modulus $E$          | $E = E_0e^{a+b}$   | 3.8   | $E_{0,\perp} = 359$ GPa, $E_{0,\parallel} = 326$ GPa |

$^a$ perpendicular and parallel to the alignment of elongated grains, corresponding to the directions of the tensile stress plane parallel and perpendicular to the grain alignment for the bending tests, see figures 40–42.

Figure 41. Porosity dependence of fracture toughness for textured porous Si$_3$N$_4$ ceramics perpendicular to the alignment of elongated grains [204]. Fracture toughness, $K_{IC}$, was calculated using the equation $K_{IC} = (2E_{\perp})^{1/2}$, where $E_{\perp}$ is the Young’s modulus of each porous sample in the plane-strain condition given by $E_{\perp} = E/(1-\nu^2)$ with Poisson ratio $\nu = 0.25$.

Figure 42. Porosity dependence of Young’s modulus for textured porous Si$_3$N$_4$ in the directions perpendicular and parallel to the alignment of elongated grains. Young’s modulus data were measured by the pulse echo method. For comparison, some data for untextured Si$_3$N$_4$ are also shown in the figure.

Figure 43. Strength damage factor as a function of quenching-temperature difference for textured and untextured porous Si$_3$N$_4$ by the water-quenching method [207]. Strength damage factor, $D_s$, is defined as $D_s = (1-\sigma_b/\sigma_{b0}) \times 100\%$, where $\sigma_b$ is the bending strength of as-received samples and $\sigma_{b0}$ is the bending strength after water-quenching.

alignment in the materials. However, there is a large difference in the $b$ value for the Young’s modulus in the directions perpendicular and parallel to the grain alignment. This is very likely to be due to the $a, b$-axis instead of the $c$-axis aligned texture, because the Young’s modulus data in the direction parallel to the grain alignment were obtained only from $a, b$-axis aligned porous Si$_3$N$_4$ prepared by partial SF using SC, in which the alignment of the $c$-axis is random. Therefore, the Young’s modulus in the direction parallel to the grain alignment should be equal to the Young’s modulus in the direction perpendicular to the $c$-axis of $\beta$-Si$_3$N$_4$, but the Young’s modulus in the direction parallel to the grain alignment should be equal to the Young’s modulus of the untextured porous Si$_3$N$_4$ materials. At the same time, the results suggest that, compared with untextured porous Si$_3$N$_4$, the excellent mechanical properties of textured porous Si$_3$N$_4$ should be primarily attributed to the alignment of elongated $\beta$-Si$_3$N$_4$ grains.

Figure 43 shows the significantly improved thermal shock resistance of textured porous Si$_3$N$_4$ in comparison with untextured porous Si$_3$N$_4$ using the water-quenching method. The critical temperature differences are $\sim 900$ and $\sim 1400$ °C for untextured and textured porous Si$_3$N$_4$, respectively. The critical temperature difference of both untextured and textured porous Si$_3$N$_4$, $\Delta T_c$, can be determined by [209]

$$\Delta T_c = \frac{\sigma_b (1-\nu)}{aE} \left(1 + \frac{Q\kappa}{ah}\right),$$  (52)
where $\sigma_b$ is the flexural strength, $E$ is the Young’s modulus, $\alpha$ is the thermal expansion coefficient, $\nu$ is the Poisson’s ratio, $Q$ is a shape factor dependent on the sample geometry ($Q = 3.25$ for a plate and $Q = 4.3$ for a rod), $\kappa$ is the thermal conductivity, $a$ is the characteristic heat-transfer length and $h$ is the surface heat-transfer coefficient. According to equation (52), the $\Delta T_c$ values of untextured and textured porous $\text{Si}_3\text{N}_4$ are 725 and 1347°C, respectively, essentially consistent with the experimental values. The slight discrepancy between the calculated and experimental values may arise from the uncertainty of the values of $h$.

8. Textured $\alpha$-Sialon

$\alpha$ and $\beta$-Sialon are two major solution ceramics of $\text{Si}_3\text{N}_4$, corresponding to $\alpha$-$\text{Si}_3\text{N}_4$ and $\beta$-$\text{Si}_3\text{N}_4$, respectively. However, compared with the elongated growth of $\beta$-Sialon grains, $\alpha$-Sialon grains normally grow into an equiaxed shape, which results in $\alpha$-Sialon ceramics having low fracture toughness and it being difficult to form a textured microstructure. However, the discovery of the preferential orientation of elongated Ca-$\alpha$-Sialon grains during HP indicates the possibility of developing textured $\alpha$-Sialon by facilitating elongated grain growth, similar to the cases of $\beta$-Sialon and $\beta$-$\text{Si}_3\text{N}_4$ [210]. The texturing behavior has also been observed in hot-pressed Ca-$\alpha$-Sialon [211], Li-$\alpha$-Sialon [212], and Nd-(+$\beta$)-Sialon [213], as clearly evidenced by the stronger relative intensities of all $(h0l)$ diffractions on the surface parallel to the HP rather than on the surface parallel to the HP.

Carman et al [214] conducted a detailed study of texturing behavior in Sm-$\alpha$-Sialon ceramics during HP and subsequent HF, as schematically illustrated in figure 44. The texture development during HP and HF was characterized by pole figure analysis, as shown in figure 45. The maximum mrd values at the edge of the pole figures (figure 45(a)) indicate that the unit cell basal plane is parallel to the HP direction (Direction 1), i.e. the elongated $\alpha$-Sialon crystals are preferentially oriented with the $c$-axis perpendicular to the HP direction. During HF, the texture development is found to depend on the forging time. When the forging time is 15 min, a unidirectional or $c$-axis oriented texture along Direction 3 is formed, as indicated by the higher maximum mrd value in Direction 3 than in Direction 1 at the edge of the pole figure (figure 45(b)). When the forging time increases to 30 min, the $c$-axis oriented texture is enhanced, as indicated by the increased maximum mrd value in Direction 3 at the edge of the pole figure (figure 45(c)) compared with the corresponding value in figure 45(b). However, with further forging (up to 60 min), the $c$-axis oriented texture along Direction 3 tends to change into a 2D oriented texture with the $c$-axis preferentially oriented perpendicular to the HP direction (Direction 2), as indicated by figure 45(d). Despite this, the $c$-axis of the $\alpha$-Sialon crystals is still preferentially oriented along Direction 3. In contrast, the 2D oriented texture in the hot-pressed sample features the random alignment of the $c$-axis of $\alpha$-Sialon crystals, as indicated by the symmetry of the pole figure (figure 45(c)). Furthermore, the authors carried out a multiple-forging experiment to further enhance the orientation of the $c$-axis as follows: a hot-pressed sample along Direction 1 was then hot-forged three times at 1800°C for 30 min—first along Direction 2, then along Direction 1, and then along Direction 2 again. Thus, a higher degree of $c$-axis orientation along Direction 3 was eventually obtained. The texture development in $\alpha$-Sialon ceramics is further indicated by the measured toughness anisotropy on the different faces using the Vickers indentation technique, as shown in figure 46. Clearly, as the $c$-axis orientation is enhanced, corresponding to 15 and 30 min forging and multiple forging, the toughness anisotropy of both Faces 1 and 2 (figure 45(b)) exhibits an increase, whereas the toughness
anisotropy of Face 3 remains almost constant. Specifically, multiple forging results in a toughness anisotropy of as high as 50%, similar to that of c-axis oriented β-Si₃N₄ prepared by tape casting (table 12). Although the α-Si₃N₄ (and α-Sialon) crystal exhibits inherent anisotropic hardness, similar to the β-Si₃N₄ (and β-Sialon) crystal [178, 215], no significant anisotropy is observed in the hardness of textured α-Sialon ceramics [214].

It has been reported that rodlike α-Sialon single crystals can be prepared by various methods, such as the sintering of powder pellets [216] and combustion synthesis [217]. Therefore, in addition to the HW method, the TGG method is anisotropy is observed in the hardness of textured α-Sialon ceramics. Several theoretical models, such as the rough–smooth plane model, the anisotropic Ostwald ripening model, the acid-base model and the differential-binding-energy model, have been proposed to interpret elongated grain growth of β-Si₃N₄, which is directly responsible for the development of the unique bimodal microstructure as well as textured microstructure. In addition, the texture development and bimodal microstructure are also related to the abnormal grain growth, which is governed by the morphology, grain-size distribution and the amount of β-Si₃N₄ nuclei.

The processing techniques for obtaining textured Si₃N₄ described in this review have been classified into two types: HW and TGG. The techniques involved in HW are HP, HF and SF, and the techniques involved in TGG are CPA, EA, TCA and SMFA based on the alignment of the β-Si₃N₄ template (or seeds). Textured β-Si₃N₄ occurs in two major structural forms: a, b-axis (or planar) orientation and c-axis orientation, depending on the processing technique. Highly c-axis oriented β-Si₃N₄ can be obtained by HF and SF using a large tensile or PSC deformation. The texture development is primarily due to the grain rotation, and the strain is the dominant factor controlling the degree of texture. In TGG, both EA and TCA are efficient for producing highly c-axis oriented β-Si₃N₄ by optimizing the aspect ratio and content of β-Si₃N₄ seeds and using HT sintering/annealing. However, the prerequisite for both orientations is a rodlike morphology and a grain size larger than the matrix grains. In contrast, the most attractive characteristic of SMFA is the independence of the particle shape of ceramics powders. Owing to the magnetic susceptibility of \( \chi_{a,b} > \chi_c \) for β-Si₃N₄, a, b-axis oriented β-Si₃N₄ is preferentially developed in an SMF. However, an RMF can be used to obtain the c-axis oriented β-Si₃N₄. It is clearly demonstrated that β-Si₃N₄ nuclei play a key role in texture development in Si₃N₄ via SMFA, and the addition of β-Si₃N₄ seeds is more efficient for increasing the degree of texture than varying sintering conditions.

The anisotropy in the hardness, elastic modulus, bending strength, fracture toughness, fracture energy, tribological and wear behavior and thermal conductivity has been clearly demonstrated and is well understood. It has been revealed both theoretically and experimentally that the thermal anisotropy is mainly determined by the intrinsic anisotropy and degree of orientation, and the effect of the rodlike grain morphology becomes less important. The anisotropic properties are maximized for the perfect c-axis orientation. Although it was reported that c-axis oriented β-Si₃N₄ exhibits higher bend and tensile creep resistance than untextured β-Si₃N₄, corresponding to the directions perpendicular and parallel to the grain orientation, respectively, the effect of the grain orientation on the bend and tensile creep resistance of textured β-Si₃N₄ is unknown. In addition, the effect of grain orientation on the erosion and contact damage resistance is also still unknown. Therefore, further studies are needed to clarify the effect of the grain orientation on the creep, erosion, contact damage and chemical corrosion behavior.

Moreover, the existence of oriented anisotropic pores allows textured porous Si₃N₄ to possess comparable and even superior mechanical and thermal-mechanical properties to textured dense Si₃N₄ and untextured dense and porous Si₃N₄. The presence of elongated α-Sialon grains allows the production of textured α-Sialon using the same methods as

![Figure 46. Toughness anisotropy of the three faces in textured α-Sialon obtained by HP and subsequent HF. The toughness was determined from the three faces shown in figure 44 by the Vickers indentation technique, as schematically illustrated in figure 29. The toughness anisotropy factor, \( f_{\text{TIA}} \), is determined by equation (43) [214].](image-url)
for producing textured $\beta$-Si$_3$N$_4$ and $\beta$-Sialon. The toughness anisotropy has been demonstrated in textured $\alpha$-Sialon prepared by HF. Because transient liquid-phase sintering enables the production of glass-free $\alpha$-Sialon ceramics, it is expected that highly $c$-axis oriented texture should lead to the further improvement in the HT properties of $\alpha$-Sialon ceramics. To establish a better understanding of the processing and properties of textured $\alpha$-Sialon, further studies are needed.

Cost and mechanical reliability are always major barriers preventing the widespread applications of Si$_3$N$_4$ ceramics. Texturing offers a unique opportunity to improve the mechanical reliability, which will help expand the applications of Si$_3$N$_4$ ceramics, particularly in extremely severe environments. From the industrial point of view, TCA is a low-cost process for producing textured Si$_3$N$_4$. However, TCA and EA have limitations for producing textured $\beta$-Si$_3$N$_4$ components with complex shapes. In contrast, SMFA is a promising process for producing textured Si$_3$N$_4$ components with complex shapes, because various colloidal forming approaches are well developed, such as slip casting, electrophoretic deposition and gel casting. However, the properties of textured $\beta$-Si$_3$N$_4$ produced by this technique are still not known. Therefore, intensive studies should be carried out on the processing and properties of textured $\beta$-Si$_3$N$_4$ by SMFA.

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