Tribological Behavior of Silicon Carbide Ceramics - A Review

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ABSTRACT

A comprehensive review on sliding and solid particle erosion wear characteristics of silicon carbide (SiC) ceramics and SiC composites is provided. Sliding or erosion wear behavior of ceramics is dependent on various material characteristics as well as test parameters. Effects of microstructural and mechanical properties of SiC ceramics are particularly focused to understand tribological performance of SiC ceramics. Results obtained between varieties of pairs of SiC ceramics indicate complexity in understanding dominant mechanisms of material removal. Wear mechanisms during sliding are mainly divided in two groups as mechanical and tribochemical. In solid particle erosion conditions, wear mechanisms of SiC ceramics are explained by elastic-plastic deformation controlled micro-fracture on the surface followed by radial-lateral crack propagation beneath the plastic zone.

Key words : Silicon carbide, Sliding wear, Erosion wear, Microstructure, Wear mechanisms

1. Introduction

SiC material has been used in industries for a long period owing to its unique combination of properties such as high hardness, high modulus, high temperature strength, good oxidation resistance, excellent wear resistance, good thermal shock resistance, and high thermal conductivity.¹⁻³⁾ The fundamental structure of covalently bonded silicon carbide is a coordinated tetrahedron, either SiC₄ or CSi₄. Although a great variety of SiC polytypes is known, important are cubic polytype referred as β -SiC, and non-cubic structures (hexagonal or rhombohedral) as α -SiC.⁴⁻¹⁰⁾ The strong covalent bond and extremely low self-diffusion coefficient ($10^{-13} - 10^{-14}$ cm²/sec) pose difficulties in densification of SiC powders at lower temperatures. The densification is possible only at extremely high pressure (30-100 MPa) and temperatures exceeding 2100°C. In order to achieve high density at lower temperatures and pressures, use of sintering aids is necessary.¹¹⁻²²⁾

Metal oxides, Al-B-C, AlN-metal oxides as sintering additives during liquid-phase sintering tailor the microstructure and improve fracture toughness of SiC ceramics.^{2,16,18,21-22)} The platelet-reinforced microstructures exhibit enhanced toughness due to the combination of intergranular crack mode, introduced by the glassy grain boundary phase, and energy dissipating processes in the crack wake.²³⁾ Sintering additive composition also affect the crystallinity of grain boundary phase of SiC ceramics. Further details on sinter-

ing and microstructure can be found elsewhere.^{15,17-19,24-28)}

Owing to attractive properties, liquid phase sintered silicon carbide (LPS-SiC) ceramics are preferred for several tribological and structural applications such as turbine parts, heat exchanger tubes, mechanical seals, bearings, cylinder liners, burner parts in fluidized bed combustion system, bullet proof vests, cutting tools etc.²⁸⁻³⁶⁾ A thorough understanding on tribological behavior of SiC ceramics is required for their successful application as triboelements. Accordingly, extensive research has been carried out towards estimating tribological potential of Si and SiC-based composites.³⁷⁻⁵⁰⁾ Available literature on tribology of SiC ceramics can be broadly divided into two categories: sliding and erosion.

Based on the conditions of sliding and characteristics of SiC and SiC composites, the coefficient of friction (COF) in unlubricated sliding studies reported to be varied in the wide range of 0.2 - 0.8, which decreased up to 0.02 in lubricating media like water, paraffin oil etc.⁴⁸⁻⁵⁷⁾ Wear rates in sliding conditions varied in the order of $10^{-7} - 10^{-4}$ mm³/N.m.^{38-44,58-59)} Erosion of SiC ceramics is mainly studied in solid particle erosion conditions, while that in cavitation conditions⁶⁰⁾ is limited. Solid particle erosion of SiC and SiC-based composites is reported to occur generally by brittle fracture as a result of lateral and radial cracking. Erosion rate ranges from 10^{-1} to 10^3 mm³/kg based on erosion parameters, and microstructural and mechanical characteristics of SiC ceramics.⁶¹⁻⁶⁷⁾ The complex tribological behavior of SiC ceramics and their composites in sliding and solid particle erosion conditions was mostly explained by several material removal mechanisms such as mechanical fracture, chipping, plastic deformation, ploughing, and/or formation and removal of tribooxide layer.^{39-48, 68)}

Considering the importance of the influence of microstruc-

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ture and mechanical properties of SiC ceramics in different wear conditions, a comprehensive understanding on sliding and solid particle erosion wear behavior of SiC ceramics and composites is provided in the present review. Effects of material parameters on wear behavior are extensively reviewed. The other important part of the review is focused on understanding dominant mechanisms of material removal as function of microstructure and mechanical characteristics of SiC ceramics in the sliding and erosion conditions.

2. Sliding Wear of SiC Ceramics

Sliding wear of SiC ceramics are reported mainly using ball-on-disk or pin-on-disk tribometer. SiC ceramics showed complex behavior of wear influenced by sliding test, material or environmental parameters in lubricated/unlubricated sliding conditions.^{39-54,68} In this section effect of microstructure and mechanical characteristics on sliding wear of SiC ceramics are separately discussed. This is followed by detailed understanding on dominant wear mechanism of SiC ceramics in sliding conditions.

2.1. Effect of microstructure

Lopez *et al.*⁴⁸ studied the sliding wear behavior of LPS-SiC ceramics with different microstructures like elongated SiC grains (*in situ* toughened LPS SiC), coarse equiaxed SiC grains, and fine equiaxed SiC grains. Sliding wear resistance of elongated grain SiC ceramics was better than that of equiaxed SiC ceramics owing to a hard interlocking network of *in situ* toughened LPS SiC grains (see Fig. 1). The isolated nature of the yttrium aluminum garnet (YAG) second phase also played a role in enhancing wear resistance of *in situ* toughened LPS-SiC. This was in contrast to the equiaxed SiC ceramics, where the grains were embedded within

a continuous YAG phase matrix.

In sliding of SiC ceramics prepared with YAG ($\text{Y}_2\text{O}_3 : \text{Al}_2\text{O}_3 = 3 : 5$), Lopez *et al.*⁵⁰ suggested reduced content of the intergranular phase or grain size for improvement in wear resistance of LPS α -SiC ceramics. Reduced content of intergranular phase and the grain refinement resulted in reduction of rate of transition from deformation-controlled wear to fracture-controlled wear and hence in wear loss.

A simple model by Lawn *et al.*⁶⁹ provided a framework for estimating the sliding wear properties in polycrystalline ceramics. The proposed model states that the plastic-deformation damage accumulates within the grains during deformation-controlled wear. This damage is in the form of dislocation pile-ups, confirmed using transmission electron microscopy.⁶⁹⁻⁷¹

It was earlier concluded that SiC ceramics sintered using 3 wt% $\text{AlN}-\text{Y}_2\text{O}_3$ (SCY3) additives revealed mostly large equiaxed grains with amorphous grain boundary phase, while SiC ceramics prepared using 3 wt% $\text{AlN}-\text{Sc}_2\text{O}_3$ (SCSc3) additives exhibited duplex microstructure of elongated and fine equiaxed grains with clean grain boundary phase (see Figs. 2(a) and (b)). SiC ceramics prepared using 3 wt% $\text{AlN}-\text{Sc}_2\text{O}_3$ additives resulted $\sim 18\%$ less wear at 13 N load. Worn surfaces of SCY3 and SCSc3 are shown in Figs. 2(c), and (d), respectively.²

Murthy *et al.*⁷² observed that SiC ceramics doped with different p-type (Al, Mg) or n-type (P) doping elements had influence on the wear behavior in two ways: (i) by directly affecting thermodynamics of tribochemical reactions or (ii) by modified frequency of grain boundaries having same orientation angle and change in the kinetics of tribochemical reactions. It was found that doping elements partly made a solid solution with SiC and rest were segregated along grain boundaries. Segregated Al reduced solubility of silica largely in water, whereas P and Mg reduced solubility to a lesser extent.⁷³ Kinetics of formation of hydrated silica var-

Fig. 1. Typical SEM images of SiC ceramics (a, b) unworn and (c, d) after sliding against Si_3N_4 ball. Note the presence of elongated grains in (b) and equiaxed grains in (a), and comparatively less wear in (d) than in (c). Other conditions of sintering and sliding wear can be found in.⁴⁸

Fig. 2. TEM images of SiC ceramics revealing (a) amorphous grain boundary in SiC ceramics prepared with 3 wt% ($\text{AlN}-\text{Y}_2\text{O}_3$) additive (b) clean grain boundary in SiC ceramics prepared with 3 wt% ($\text{AlN}-\text{Sc}_2\text{O}_3$). Worn surfaces of (a) and (b) are respectively shown in (c) and (d).²

ied with solubility and hence led to change in frictional characteristics of SiC ceramics.

For improvements in the sliding-wear resistance of pressureless LPS-SiC under diesel lubrication, Ciudad *et al.*⁷⁴⁾ suggested (i) reduction in intergranular phase (second phase) content, and (ii) decrease in grain size or increase in grain aspect ratio. They reported that SiC ceramics with similar grain size (of 1.1 μm) but having large amount of second phase exhibited high wear. Second phase led to decrease in hardness and increase in wear.

Ortiz *et al.*⁵⁴⁾ proposed a colloidal processing route to produce dense SiC ceramics via pressureless liquid-phase sintering (PLPS). In addition to uniform mixing, the colloidal processing route resulted in a nano film of ~ 5 nm thickness on the surface of SiC particles. Compared to conventional mixing and sintering, complete densification achieved in shorter sintering time and minimum defects found in ceramics sintered using colloidal processed powders. The time required for the transition from mild to severe wear was more than double for SiC ceramics prepared by colloidal processing as compared to that for the SiC ceramics prepared by conventional processing. It was advocated that, though the microstructures were similar in terms of the size and morphology of grains for ceramics prepared through both processing routes, the clear interphase with fully dense structure led to a superior wear resistance.

Zhou *et al.*⁷⁵⁾ reported that silicon carbide-graphite (SiC-C) composites exhibited lower coefficients of friction during sliding than the monolithic SiC. The flake like graphite particles present in SiC-C composites (see Fig. 3) were expected to provide solid lubrication with easy shear during sliding. However, in the high contact stress condition (pin-on-disk tests), wear rates enhanced sharply from 0.38 to 3.90×10^{-6} $\text{mm}^3/\text{N.m}$ for SiC-C composites which was attributed to the fracture of graphite flakes despite lowered coefficients of

Fig. 3. Fracture surfaces of (a) SiC and (b) SiC- 20 vol% graphite (SiC-C) composite. The graphite particles in SiC-C composite remain flake like, as indicated by the arrows.⁷⁵⁾

friction when compared to low contact stress condition (block-on-ring test) (See Table 1).

In other research,⁷⁶⁾ SiC ceramics with 0, 10, 30 or 50 wt% WC were slid against SiC ball. WC reinforcement led to a decrease in size of equiaxed grains from 835 μm to 578 μm of SiC ceramics and resulted improvement in wear resistance from 1.3×10^{-5} $\text{mm}^3/\text{N.m}$ to 4×10^{-6} $\text{mm}^3/\text{N.m}$ (see Fig. 4).

2.2. Effect of mechanical properties

Mechanical properties such as fracture toughness, hardness, and elastic modulus have influence on sliding wear characteristics of SiC ceramics. Hardness and fracture toughness of SiC ceramics are reported to generally range from 16 GPa to 27 GPa, and 2.3 $\text{MPa.m}^{1/2}$ to 8.3 $\text{MPa.m}^{1/2}$, respectively based on types of sintering/additives, microstructure etc.^{49,51,77-80)} Particularly, the influence of hardness and fracture toughness on friction and wear behavior was debated in most of the reports.

A sharp indenter model and a blunt indenter model were proposed to estimate the sliding wear broadly in the tribocontact of brittle materials in absence of any layer formation. According to Marshall *et al.*⁸¹⁾ for the tribocontact of sharp indenters, wear in brittle materials occurs due to the formation and propagation of lateral cracks. For a given applied load (P), total sliding distance (S), hardness (H), fracture toughness (K_{Ic}) and elastic modulus (E), wear vol-

Fig. 4. Etched surfaces of SiC ceramics prepared with (a) 10% WC, and (b) 50% WC.⁶¹⁾ Sliding wear volume as function of WC content is shown in (c).⁷⁶⁾

Table 1. Friction and Wear Data for SiC-C Ceramics for Different Triboconfigurations⁷⁵⁾

Specimen	Average coefficient of friction		Wear rate ($\times 10^{-6} \text{ mm}^3/\text{N.m}$)	
	Block-on-ring	Pin-on-disk	Block specimen in block-on-ring test	Pin specimen in pin-on-disk test
SiC	0.64	0.28	4.60	0.38
SiC- 10 vol. %C	0.55	0.15	3.20	2.90
SiC- 20 vol. %C	0.49	0.14	6.20	2.30

ume of the brittle solid (V_s) can be assessed using the following equation:⁸¹⁾

$$V_s = \alpha \frac{P^{9/8}}{K_{IC}^{1/2} H^{5/8}} \left(\frac{E}{H} \right)^{4/5} S \quad (1)$$

where α is a material constant. In case of blunt indenter pressed against a brittle solid, surface ring cracks are generated, which propagate towards downwards with repeated sliding contact and develop conical cracks to cause pull-out of material.⁸²⁾ Following Hertzian analysis, when there was no tribochemical layer formation, Tewari *et al.* estimated wear volume (V_B) for brittle materials in ball-on-disk fretting wear contacts as per the following relation.⁸³⁾

$$V_B = \frac{2\pi\beta^2 s}{3} \left(\frac{3PR}{4E} \right)^{2/3} \sqrt{\left(\left(\frac{4\chi^2 EP}{3\beta^2 K_{IC}^2 R} \right) - 1 \right)} \quad (2)$$

where R is radius of the ball, χ and β are constants, and s is the stroke length in fretting contact.

SiC ceramics sintered with small amount (0.2 wt%) of Y_2O_3 additives attributed to high strength (561 MPa) and high hardness (27.7 GPa) than the SiC ceramics sintered with large amount (3 wt%) of Y_2O_3 additives (542 MPa and 26.1 GPa), and exhibited superior wear resistance.⁸⁴⁾ Li *et al.*⁸⁵⁾ found that toughened SiC prepared by hot pressing showed high hardness and fracture toughness compared to the SiC ceramics prepared through pressureless sintering or reaction sintering. The same trend reflected for resistance against friction and wear in water-lubrication conditions against SiC counterbody. Lopez *et al.*⁵¹⁾ reported that elongated-grain LPS SiC, despite having lower hardness, showed almost a half of the wear rate than that obtained for equiaxed-grain LPS SiC, and reflected less extent of dislocation plasticity. They also differentiated the extent of wear as mild wear dominated by hardness and severe wear dominated by fracture toughness of LPS SiC ceramics. Kovalcikova *et al.*⁸⁶⁾ used brittleness index (H/K_{IC}) of material for qualitative estimation of specific wear rate of SiC against Si_3N_4 . They observed that brittleness index had dominant role in the wear response. Their results showed that resistance to wear increases with increased index number. In other study on SiC-WC composites,⁷⁶⁾ the influence of hardness and fracture toughness on the sliding wear of the SiC-WC composites was realized. The maximum hardness of 26.3 GPa for SiC reinforced with 30 wt% WC particles resulted in homogeneous dispersion of WC particles in the SiC matrix, whereas, a higher fracture toughness obtained due to crack deflection and bridging by a large amount (30 or 50 wt%) of WC particles.⁶¹⁾ The minimum material removal for SiC ceramics sintered with 50 wt% WC attributed to the maximum fracture toughness. Thus, fracture toughness appeared to carry important role than hardness towards estimating the extent of sliding wear for the investigated SiC-WC composites.

Llorente *et al.*⁷⁹⁾ studied tribological nature of SiC compos-

ites prepared with different types and amounts of graphene-based sources. They demonstrated the importance of graphene fillers in resisting wear during sliding. According to them, the choice of the most suitable amount and type of graphene in SiC composites for tribological and mechanical applications is dependent on their specific working requirements. SiC composites with 20 vol% graphene nanoplatelets (GNPs) clearly exhibited 72% higher wear resistance compare to the monolithic SiC (see Fig. 5). It was demonstrated that an optimum combination of mechanical properties is required to provide the best wear resistance for SiC ceramics.

Amirthan *et al.*⁸⁷⁾ studied the reciprocating sliding friction and wear characteristics of cotton fabric/teak wood/jute fiber based Si/SiC. Among the investigated composites, fine teak wood and cotton based samples exhibited less COF of 0.25 and 0.23 and less wear rate of 3.0×10^{-6} and 1.24×10^{-6} mm³/N.m respectively in dry sliding condition. This was attributed to the carbon film formation that acted as lubricating media on the wear track. Alumina particles, generated as debris from alumina counterbody balls acted as third body for harder surfaces. Third body abrasion of the surfaces led to high wear of harder chemical vapor infiltration treated cotton based sample. On the other hand, porous nature of the coarse wood particle and jute fiber based sample surfaces exhibited high wear.

Candelario *et al.*⁸⁸⁾ introduced graphite nanodispersoids into the microstructure of a fine-grained LPS SiC ceramics fabricated by spark plasma sintering and studied the sliding wear characteristics in regular diesel fuel (viscosity of ~ 3.8 cSt) lubricating conditions. Addition of graphite nanodispersoids was detrimental to mild wear and transition from mild to severe wear occurred in shorter time due to softer graphite particles. On other side, the addition of graphite nanodispersoids led to increased fracture toughness and provided external lubrication after grain pull-out. This was resulted in effectively decreasing wear damage for

Fig. 5. The wear rate and fracture toughness of SiC- graphene nanoplatelets composites as function of content of graphene nanoplatelets.⁷⁹⁾

Table 2. Summary of Major Findings from Sliding Wear Studies of SiC Based Ceramics

Materials (SiC based ceramics/ Counter body)	Sliding test parameters			Major Findings	References
	Speed	Load	Environment or Lubrication		
SiC pin/SiC disk	2 mm/s	5, 30, 50, 100 N,	50±2% RH and 19±2°C	A hydrated silicium oxide film forms and functions as a lubricating film to reduce friction and wear.	89)
SiC ball/SiC disk	10mm/s	250 mN	vacuum	In vacuum, the transition from severe to mild wear was connected with smoothening of surface roughness.	41)
	1 m/s	1N	Air		
Si ₃ N ₄ ball/LPS-SiC disks	~ 0.04 m/s to ~ 0.01 m/s	240 N to 100 N	Diesel (viscosity ~ 3.8 cst)	Resistance to sliding decreases with increasing contact load and sliding speed.	74)
Si ₃ N ₄ ball/LPS-α-SiC disk	~ 0.04 m/s	70 N	Paraffin oil (viscosity ~ 34 cst)	The superior sliding wear resistance of low-cost <i>in situ</i> toughened SiC ceramic was most likely due to the formation of interlocking networks of elongated SiC grains.	50)
SiC ball/SiC-3 wt% (AlN-Sc ₂ O ₃ or AlN-Y ₂ O ₃) disk	0.21 m/s	1 N, 6 N or 13 N	20 -580 °C and 50 - 10% RH	SiC ceramics sintered with 3 wt% AlN-Sc ₂ O ₃ additives exhibited less wear. Surface grooving and microcracking occurred at low load (1 N). Tribochemical wear was dominant at 6 and 13 N loads for all the ceramics.	2)
				At higher humidity (60% RH), the friction coefficient value (0.25) is almost independent of doping elements.	
α-SiC ball/ Al, Mg or P-doped SiC	36 mm/s	1 N,	30% - 60% RH	At lower humidity (30% RH), the effect of doping elements (Al, Mg and P) is more pronounced.	42)
				As the humidity increases from 30% to 60%, the nature of damage changes from mechanical to tribochemical.	
SiC-MoSi ₂ / Alumina cylinder	0.5, 1.0 and 2.0 m/s	15, 30 and 50 N	20 - 25°C and 35 - 55% RH	Mild to severe wear transition with speed (high wear at high speed of 2 m/s)	90)
SiC _f -SiC pin/hardened steel disc	~ 100 to ~ 850 rpm	11 N	Dry Atmosphere	SiC _f /SiC composite with carbon interface shows high hardness and least wear rate. SiC _f /SiC composite with BN interface, subjected to an intermediate heat treatment showed lower hardness and higher wear rate	59)
SiC or Al ₂ O ₃ balls/ Pressure less sintered SiC disk	Stroke 100 - 1600 μm, frequency 2.5 - 20 Hz	1-10 N	1 to 100% RH	Tribo-oxidation was determined as the main wear mechanism.	91)
SiC cylinder/SiC plate	Stroke 8mm, frequency 10 Hz	60 N	Isooctane and distilled water	The self-mated ceramics showed greater values of friction coefficient in isooctane than in distilled water	60)

a prolonged sliding. Important findings from sliding wear studies of SiC based ceramics are listed in Table 2.

2.3. Mechanisms of material removal in sliding wear

In this section, studies relating to material removal mechanisms during sliding wear are discussed in terms of material characteristics. Different results for frictional behavior between varieties of pairs of SiC ceramics during wear indicate complexity in frictional mechanism. Kato⁹²⁾ divided wear mechanisms of SiC ceramics in sliding wear conditions into two modes: (a) mechanical wear, where various types of

cracks like lateral, median and radial cracks are generated by friction around the Hertzian contact zone (see Fig.6), and (b) tribo-chemical wear, where silicon oxide was found on the surface in unlubricated sliding wear.^{93,94)}

Wang *et al.*⁴³⁾ found that tribological contact stresses predominantly affect mechanisms of material removal. They found that SiC ceramics removed via plastic deformation induced microfracture for less stress on contact surface. As the contact stress increased to a critical level, varieties of cracks like partial cone cracks, lateral/shallow cracks, and radial cracks etc. are generated. Further increase in contact stress propagates cracks to intersect each other and detach

Fig. 6. Various types of cracks induced by friction in unlubricated sliding wear conditions.⁹³⁾ (where W is load and F is frictional force).

chunks of material. The detached material crushed into fine particles by further tribological contact and evolved as wear debris. The dominant damage patterns with contact stress severity for indentation, scratch, and wear tests are synchronized in Table 3.

Lopez *et al.*⁵¹⁾ reported that for equiaxed-grained SiC ceramics, sliding wear initiated with plastic dislocation as mild wear and led to sever wear, with fracture and pull-out of material.^{37,69,85,95-96)} For coarser microstructures, transition in wear mechanism occurred sooner and exhibited poor wear resistance.^{43,47,50,69-70,96-99)} However, these mechanisms were not observed for highly heterogeneous ceramics such as elongated-grain LPS-SiC.⁵¹⁾ For elongated grain LPS SiC ceramics, despite the grain-boundary fracture, interlocking network was responsible for the surface integrity of worn surface which delayed transition in wear mechanism.

Kovalcikova *et al.*⁸⁶⁾ studied wear mechanisms for SiC ceramics slid against different counterparts Si_3N_4 , Al_2O_3 , WC-Co and ZrO_2 materials. They found microfracture and removal of oxides against any counterpart. But the extent of material removal was different. The steady state coefficient of friction and specific wear rate were affected with different tribological partners. The SiC/ Si_3N_4 couple exhibited the

highest friction probably due to chemical similarity of counterparts.

Andersson *et al.*⁴⁵⁾ reported two types of oxidation in sliding at elevated contact temperatures: passive oxidation and active oxidation. At elevated temperature (usually above 575°C), SiC oxidizes to SiO_2 according to the following reactions:¹⁰⁰⁻¹⁰¹⁾

Above reactions occur at an appropriate pressure in oxygen environment. If temperature is above 1000 °C, with low partial pressures of oxygen, the following reactions might occur on SiC ceramic surface.¹⁰¹⁾

The SiO powder formed was brownish to yellowish in color and often formed with β -SiC powders in an argon thermal plasma jet.¹⁰²⁾

Under passive oxidation layer, loose wear debris consisting of SiO_2 formed a protective and less shear strength tribofilms which further favored to reduced wear rates. Against passive oxidation layer, wear rates varied from 10^{-6} to $10^{-5} \text{ mm}^3/\text{N.m}$ and the coefficients of friction was approximately 0.5. As SiO did not form tribofilms and favored to active oxidation to some extent, wear rates increased to 10^{-5} – $10^{-4} \text{ mm}^3/\text{N.m}$ and the coefficients of friction to 0.8. It was found that the additive composition had minimal effect on frictional behavior of SiC ceramics under dry sliding conditions when compared to sliding load.²⁾ Wear mechanisms changed from surface grooving and micro-cracking at lower load (1 N) to dominant tribochemical wear at higher (6 or 13 N) load for all the ceramics. In other study,⁸⁴⁾ sliding wear behavior of SiC ceramics sintered with (0.2 wt.% or 3 wt.%) Y_2O_3 additive, microcracks induced fracture and pull-out were found responsible for the material removal.

In self-mated sliding of SiC ceramics, Gahr *et al.*⁴¹⁾ reported that running-in period was much longer in vacuum than in humid air and would favor tribochemical smoothening of surface roughness. At a tribocontact of relatively

Table 3. Contact Stress Severity Against Damage Patterns for SiC Ceramics⁴³⁾

Contact Stress	Dominant damage patterns in sliding of SiC ceramics		
	Indentation	Scratch	Wear
Small (below the elasticity limit)	No damage (Elastic Deformation)	No damage (Elastic Deformation)	No damage (Elastic Deformation)
Moderate (below the plastic limit)	Plastic deformation	Plastic deformation	Plastic deformation controlled wear
Large (near the critical failure stress)	Cone/Ring cracks	Partial Cone/Ring cracks	Microcrack controlled wear or grain pull-out
Very large (exceeds the critical stress)	Radial crack Lateral crack	Microfracture caused by the intersection of radial, lateral and partial cone cracks	Microfractured-controlled wear

smooth surface, formation of thin water film led to increased friction owing to menisci formed at asperities. On the other hand, thick water films separated the mated bodies and resulted for reduced friction due to low viscosity of H_2O and assumed a proper wettability of tribosurfaces with water film, i.e. hydrophilic surfaces.

Available data on the effect of humidity on coefficient of friction and wear rate of self-mated SiC ceramics is compiled and presented as Fig. 7.^{72, 85, 91, 93} Murthy *et al.*⁷² found that at lower humidity, major wear mechanism was fracture of tribosurface followed by attrition, while tribochemical reaction (oxidation or hydrolysis) was a leading wear mechanism at higher humidity. Owing to the availability of more water molecules, oxidation ($SiC + 2H_2O \rightarrow SiO_2 + CH_4$) and hydrolysis ($SiO_2 + H_2O \rightarrow Si(OH)_4$) reactions were accelerated at higher humidity. The faster kinetics of tribochemical reactions resulted in decreased wear with a narrow difference in friction.

Li *et al.*⁸⁵ observed difference in wear mechanism during sliding in water lubrication conditions for SiC ceramics prepared through different processing routes. The wear occurred by pull-out of grains, which was mainly controlled by free Si along grain boundary for the reaction sintered SiC. Tribochemical oxidation in water and local grain fracture occurred due to its low toughness for the pressureless sintered SiC. Excellent tribological properties and low friction were attributed for the hydrodynamic lubrication for the toughened SiC. It was believed that tribochemical oxidation took place and an ultraflat surface consisting of SiO_2 and $(SiO_2 \cdot nH_2O)$ formed on the worn track.

Candelaria *et al.*⁸⁸ found less wear of SiC with graphite nanodispersoids compared to SiC ceramics, due to the combination of (i) higher toughness, which hinders the crack propagation and coalescence responsible for the grain pull-out, (ii) lower hardness of the wear debris due to its lower SiC concentration, and (iii) external lubrication imposed by pull-out of graphite nanoparticles, which reduced abrasion by third bodies. Zhu *et al.*¹⁰³ studied polishing of SiC ceram-

ics in oxidant solution of 3 wt% CrO_3 . The removal of material by a chemical dissolution, stimulated by friction was referred as tribochemical.

Ciudad *et al.*⁷⁴ studied wear mechanisms in diesel for pressureless LPS-SiC. They found a two-step wear process, initially controlled by plastic deformation (corresponding to mild wear regime) followed by fracture of materials (corresponding to severe wear regime). Major characteristics of surfaces in mild wear and severe wear regimes are schematically illustrated in Fig. 8. Erickson *et al.*¹⁰⁴ observed chemical wear on mated surface as major deterioration mechanism in oil/water lubricated sliding conditions along with several mechanisms like plastic deformation, subsurface microcrack formation, microfracture and microabrasion of the surface of worn SiC. The absence of tribofilm on the contact surface indicates that wear debris transported out of the system or directly dissolved by the oil/water lubricant as water is able to dissolve SiO_2 easily.¹⁰⁵⁻¹⁰⁷

Zhou *et al.*⁷⁵ reported that the tribochemical reaction led to material removal for monolithic SiC and SiC-C (graphite) composite. For SiC, they found rough worn surfaces resulted in fracture and three body abrasion, whereas graphite particles exerted a lubricating effect and resulted lower coefficients of friction for the SiC-C composites. Micele *et al.*⁹⁰ studied the tribological characteristics of SiC-MoSi₂ composites against alumina in dry sliding. The material loss was caused by ploughing effects, while no significant effect of brittle fracture observed on the worn surface of the ceramic composite.

Recently, sliding wear mechanisms were studied for monolithic SiC and SiC-WC composites against SiC counterbody.⁷⁶ For monolithic SiC ceramics, blunt indenter-induced mechanical fracture with micro-cracking was found as dominant material removal mechanism, while WC addition reduced grain pull-out for SiC-WC composites leading to reduced wear rates. Analysis of debris collected from worn surfaces indicated increase in compaction or oxidation with WC addition. Wasche and Klaffke¹⁰⁸ studied the wear mechanism for SiC-TiC-TiB₂ in water when slid against SiC or Al_2O_3 . In case of SiC/SiC and Al_2O_3 /SiC, no films were formed. An oxide film was formed during wear of SiC-TiC and SiC-TiC-TiB₂ materials. Reduction in wear of materials was affected by formation of titanium oxide at the tribocontact. In case of Al_2O_3 ball and composite disk system, the tribooxidatively formed layer was partly adhered on to the ball. The wear of SiC-TiC or SiC-TiC-TiB₂ disk against

Fig. 7. Average steady state coefficient of friction and wear rate of SiC ceramic with respect to humidity against SiC counterbody.^{72, 85, 91 and 93}

Fig. 8. Schematic illustrations of worn surface characteristics of SiC ceramics after sliding in diesel in (a) mild wear and (b) severe wear regimes.⁷¹

SiC or Al₂O₃ ball was reduced by a factor of 10 compared to monolithic SiC disk system.

3. Erosion Wear of SiC Ceramics

Erosion can be defined as removal of material from a surface due to interaction between the surface and a fluid, or impinging liquid or solid particles.^{65,109-110} Considerable research has been carried out in understanding solid particle erosion wear behavior of SiC ceramics and their composites.^{62,65,67,109-117}

3.1. Effect of microstructure and mechanical properties

Solid particle erosion of SiC ceramics and their composites is reported to occur generally by brittle fracture as a result of lateral and radial cracking.^{62,67,110-113} Bell and Rogers¹¹⁸ noted fracture toughness as predominant factor in estimating erosion wear characteristics of brittle materials at normal impact of erodent,^{65,119} whereas hardness was found important at low impingement angles. Shetty *et al.*¹²⁰ also reported hardness as predominant property of reaction-bonded SiC ceramics to estimate erosion behavior of ceramics. Routbort and Matzke¹¹⁴ studied the erosion for SiC ceramics prepared through hot-pressing and reaction-bonded sintering techniques. The Vickers hardness (H) and fracture toughness (K_{IC}) of these ceramics were used to estimate the erosion rate (ΔW) of ceramics using following equation:

$$\Delta W = C H^a K_{IC}^b \quad (7)$$

where, C is proportionality constant and the exponents a and b valued according to details of the used erosion model¹¹⁴ (see Table 4). However, the wear rates estimated using existing wear models were not in agreement with the experimentally measured wear rates. This can be primarily attributed to the fact that these models neglected important contributions from microstructure of the target material in estimating erosion loss.^{65-67,112-114}

Wang *et al.*⁶⁵ studied the effect of microstructure on erosive wear of SiC ceramics, prepared by pressureless sintering, hot pressing and hot isostatic pressing techniques. Microcracking with some extent of plastic deformation was observed as main material removal mechanism at higher

impact angles.¹²¹⁻¹²² Grain refinement and reduction in the amount and size of pores of hot isostatically pressed SiC resulted in superior wear resistance.

SiC ceramics sintered with 10, 30 or 50 wt% WC particles via hot press technique were subjected to erosion against SiC and alumina erodent.⁶¹ Compared to erosion rates against alumina erodent, erosion rates against SiC erodent increased by two orders (from 10 to 10³ mm³/kg) for normal impingement (see Table 5). SiC ceramics achieved steady state erosion after 5 min of erosion against SiC erodent which delayed to 20 min against alumina. Maximum erosion resistance was obtained for SiC ceramics sintered with 30 wt% WC which possessed maximum hardness. SiC ceramics sintered with 50 wt% WC ceramics having highest fracture toughness showed lower wear resistance than SiC ceramics sintered with 30 wt% WC (see Fig. 9). This implies that hardness played important role in erosion of SiC-WC composites.⁶¹

Table 4. Values of (H^a K_{IC}^b) Calculated for the Various Steady-state Erosion Models and Experimentally Measured Erosion Rates for SiC Ceramics.¹¹⁴ HP-SiC and RB-SiC respectively Indicate Hot Pressed SiC And Reaction bonded SiC, while H and K_{IC} Indicate Hardness and Fracture Toughness, Respectively

Materials	H ^{0.11} K _{IC} ^{-1.3}	H ^{-0.25} K _{IC} ^{-1.3}	H ^{0.48} K _{IC} ^{-1.9}	Measured Erosion Rate (× 10 ⁻³ g/g)
HP-SiC	0.54	0.01	11.10	2.00
RB-SiC	1.34	0.03	40.00	12.00
RB-SiC + 0.2 vol% Si	0.73	0.02	16.20	10.00
RB-SiC + 0.13 vol% Si	0.80	0.02	18.00	4.50

Table 5. Average Erosion Rates of SiC-WC Composites at Normal Impact of Alumina and Silicon Carbide Erodents⁶¹

Material	Average erosion rate (mm ³ /kg)	
	Alumina erodent	Silicon carbide erodent
SiC ceramics	18.31	1044.63
SiC- 30 wt% WC	6.57	850.73
SiC- 50 wt% WC	20.49	1223.24

Fig. 9. Eroded surfaces of (a) SiC (b) SiC-30 wt % WC and (c) SiC-50 wt % WC composites at normal impact of SiC erodent.⁶¹ Hardness data for the respective ceramics is also shown.

Wang and Levy¹²¹⁾ studied the erosion behavior of SiC fiber reinforced SiC composite at 25°C and 850°C and found an inverse relation of density and hardness to the erosion rate of the composite. The erosion rate reduced by one order of magnitude at 850°C (0.10 mg/g) compared to that at 25°C (2.20 mg/g). At 850°C, the difference between erosion rates at 90° and 30° (0.02 mg/g) was much less than that found at 25°C (1 mg/g), which attributed to increased ductility of composite at 850°C. Similarly in a recent study,⁶³⁾ solid particle erosion rate of SiC-Si₃N₄ composite at elevated temperature increased firstly up to 800°C, and then decreased with increase in temperature to 1400°C. At lower temperature (up to 800°C), wear occurred by brittle fracture where stripping of aggregates of composite resulted in loss of protection from the matrix. On the other hand, plastic deformation and oxidation protection dominated at higher temperature (beyond 800°C). The highest erosion rate at 800°C for SiC-Si₃N₄ is contradictory to the lowest erosion at 850°C for SiC ceramics.⁶⁴⁾

Amirthan *et al.*¹²⁴⁾ found cleavage like brittle fracture on biomorphic Si/SiC composite surface exposed to erosion. Fine teak wood particle based Si/SiC composite showed least erosion compared to other Si/SiC composite. Further,

coarse teak wood particle based Si/SiC composite with large pores caused the highest erosion rate. Lopez *et al.*¹²⁵⁾ studied the erosion behavior of biomorphic (eucalyptus and pine reinforced) SiC, reaction-bonded SiC, and hot-pressed SiC and found that erosion resistance was highest for biomorphic based SiC ceramics. Owing to the highest hardness, hot-pressed SiC exhibited least erosion rate.

3.2. Mechanisms of material removal in erosion wear

The dominant mechanisms of material removal in solid particle erosion wear conditions of brittle ceramics are elastic-plastic deformation based micro-fracture, and crack formation below the plastic zone of subsurface.^{62,67,110-112)} Major findings from studies on erosion wear mechanisms for SiC ceramics are listed in Table 6.

A detailed investigation of impact sites of worn surface in brittle materials exhibits an intense plastic deformation beneath the immediate area of the contact.¹²⁶⁻¹²⁹⁾ The elastic-plastic zone at the impact site is the major driving force for the surface fracture that led to material loss on erosion.¹³⁰⁻¹³²⁾ The residual stresses generated under plastic zone on impact of erodent to the brittle materials originate small lateral cracks to grow beneath the impact sites. Initially

Table 6. Summary of Major Findings from Studies on Erosion Wear Mechanism for SiC Based Ceramics

Materials (SiC based ceramics/ erodent)	Erosion test parameters			Major findings	References
	Particle velocity & feed rate	Impact angle	Environment		
Reaction Bonded SiC/ Angular Al ₂ O ₃	Velocity varied from 53.8 m/s to 150.7 m/s	Impact Angle Varied from 10° to 90°	Normal atmosphere	The steady state erosion rate was lower for smallest erodent particle. With leading brittle fracture, some plasticity also observed in the SEM micrographs.	67)
Biomorphic SiC, Reaction-bonded SiC, and hot-pressed SiC/ Angular SiC	Particle velocity = 100 m/s, feed rate of 8 gm/min	Normal Impact	Vacuum (approx. 500 mTorr),	Both lateral and radial cracks formed for biomorphic SiC ceramics, whereas reaction-bonded SiC eroded by formation and propagation of only lateral cracks. Hardest hot-pressed SiC ceramic found as the most erosion resistant.	122)
α-SiC target specimens/SiC	Velocity 35, 65 and 90 m/s, feed rate of 10 g/min,	Impingement angles of 15, 45, 60 and 90°,	Ambient Temperature	Reduced amount and size of pores, and reduced grain size led to improved wear resistance. Microcracking and some extent of plastic deformation was found as dominant material removal mechanism is mainly with some extent of plastic behaviour of materials.	65)
Hot-pressed monolithic SiC and SiC-TiB ₂ / Angular SiC	Velocity 40–100 m/s, feed rate: 0±0.5 gm/min,	Impact angles: 30–90°	Room Temperature	The erosion rate was lower for monolithic SiC with higher hardness value than SiC-TiB ₂ composite of higher fracture toughness. Erosion of these materials was controlled significantly by the plastic deformation.	62)
SiC _f -SiC / Silica sand	Particle velocity: 32 m/s and 44 m/s, feed rate: 1 kg/min,	Impinge angle: 15°, 30°, 45°, 60°, 90°	Ambient Temperature	Higher hardness of SiC _f -SiC composite with carbon interface led to least erosion. Composite with boron nitride interface showed lower hardness and higher erosion loss.	59)

Fig. 10. Schematic representation of erosion mechanisms for SiC-Si₃N₄ composites at different temperatures.⁶³⁾

these cracks propagate parallel to the impacted zone of target surface and further interact to the surface resulting in material loss. Wang and Mao¹³³⁾ observed that size and type of cracks generated on the eroded surface are dependent on several factors like erodent shape and size, mass flow rate, velocity and angle of impingement of erodent, temperature and hardness, toughness and microstructure of target material.^{61, 65, 113-114)}

Two types of crack systems are generally found in erosion at low velocities; (a) cone cracks generated due to impact of blunt (rounded) erodent^{113,134-135)} and (b) lateral and median cracks generated due to impact of sharp (angular) erodent angular.^{113,126,136-137)} In case of high velocities of erodent, material is removed by plowing, severe cracking and chipping.^{110,113,130,132,134)} In other study,⁶⁷⁾ precise mechanisms for the erosion of reaction bonded-SiC were not clear, but reduced erosion wear rate of mixed phase SiC-Si region was related to either by arresting of cracks around the small SiC grains or by increased apparent volume fraction of uncovered erosion resistant fine SiC grains of the two-phase region, generated beneath the impacted sites.

Suh *et al.*¹³⁷⁾ studied the effect of SiC fiber reinforcement on erosion wear of SiC ceramics. During erosion of the composite, initially SiC fibers were detached from the interface and matrix of SiC subsequently eroded out. Pores presented in matrix with aperture around the fibers led to easy removal of material from the impact sites. The erosion wear was also attributed to issues regarding durability and reliability that depend on geometric discordance, irregular performance, and related energy dissipation.

Gochnour *et al.*¹³⁸⁾ studied erosion behavior of SiC-Al₂O₃ composites against alumina erodent. They found surface appeared smooth and polished, with grain pull-out regions in erosion of SiC-5wt % Al₂O₃ composites. Softer Al₂O₃ phases resulted in grain pull-out regions and smooth zones were corresponded to wear of harder SiC phase. However, severe grain pull-out occurred in erosion of SiC-50 wt % Al₂O₃ composite and resulted in isolated smooth regions only. The steady-state erosion rate (ΔE) for brittle materials was proportional to erodent size and found to be related as

$$\Delta E \propto R^{2/3} \quad (8)$$

where R is the erodent radius.^{113,139)} This was predicted by both the dynamic¹¹³⁾ and the quasi-static¹³⁹⁾ theories of erosion for brittle materials. However, Kim and Park¹⁴⁰⁾ found

that erosion rates of SiC and SiC-TiB₂ were not increased monotonically with the increased erodent size, in contrast to the prediction by theories. The eroded surface of SiC exhibited deformed lips and cracks, whereas intergranular fracture observed together with deformed lips for SiC-TiB₂ composite.

Limited reports are available on erosion behavior of SiC ceramics and composites as function of temperature. Wang and Levy⁶⁴⁾ reported that erosion for SiC fiber reinforced SiC composite at 25°C occurred by cracking and chipping of the matrix as well as the fibers, while erosion rates considerably decreased at 850 °C owing to the increased ductility. Li *et al.*⁶³⁾ studied the erosion mechanism for SiC-Si₃N₄ composites at different temperatures (25 °C-1400 °C). At low temperature, material removal occurred mainly by brittle fracture and stripping of aggregates as the protection from the matrix was lost. At higher temperature, erosion mechanisms were primarily ascribed to plastic deformation as well as oxidation protection. Oxidation of Si₃N₄ and SiC on the material surface at 1400 °C generated a thin dense layer of SiO₂ glassy film, which further reduced direct contact of erodent with the material surface at a certain extent and resulted in less loss of the material (illustrated schematically as Fig. 10).¹⁴¹⁾

4. Summary and Future Scope

Considering the wide range of tribological applications, wear behavior of SiC and its composites is extensively studied. In the present review, tribological characteristics of SiC ceramics and composites in sliding and solid particle erosion conditions are comprehensively discussed.

It is found during sliding that wear and friction characteristics of SiC ceramics vary with microstructural alteration. Hard interlocking network of elongated grains or high aspect ratio grains led to enhanced wear resistance. Second phases and intergranular phases also play a dominant role on wear and friction behavior. SiC ceramics with reduced weak second phase content and grain refinement exhibited improved wear resistance. Further, SiC ceramics with clear grain boundaries showed improved wear resistance compared to amorphous grain boundaries. The transition from mild to severe wear occurs for SiC ceramics with coarse grain structure than with fine grain structure. Doping elements affect kinetics of tribochemical reaction and lead to different characteristics of friction and wear. In understand-

ing the influence of mechanical properties on wear of SiC ceramics, research is mainly debated on the dominant effect of fracture toughness and hardness. It is widely accepted that crack deflection or crack bridging by reinforced phase led to less extent of material removal during sliding wear of SiC composites. Furthermore, the ratio of hardness and fracture toughness (H/K_{IC}), known as brittleness index is found to give qualitative estimation of wear for SiC ceramics. Superior resistance to wear was found for ceramics with high brittleness index. Reinforced phases like graphene nanodispersoids led to an increase in fracture toughness and reduced hardness, the combination of which reduced the mild to severe wear transition.

Sliding wear mechanisms of SiC ceramics and composites can be categorized into two groups: mechanical and tribochemical. In mechanical wear, cracks and abrasion dominate the worn surfaces of SiC ceramics, whereas oxides were found on the tribosurface in tribochemical wear. With regards to contact stress conditions, wear occurred through plastic deformation induced microfracture in low contact stress, whereas cracks induced fracture led to material removal in high contact stress. High friction is found against the counterbody having chemical similarity with SiC ceramics, whereas composition of sintering additives had minimal effect on frictional behavior of SiC ceramics. At elevated sliding temperature, two types of oxidation: passive oxidation with reduced wear and active oxidation with increased wear are reported for SiC ceramics. Running-in period is reported to be higher in vacuum than in humid air for SiC ceramics. Tribochemical smooth layer formed in humid air led to less friction. During oil/diesel fuel lubrication, wear occurred in two steps: initially controlled by plastic deformation and then followed by subsurface microcrack formation, microfracture and micro-abrasion.

Solid particle erosion rate is found to be higher whereas steady states attained earlier against harder erodent particles. Erosion resistance is improved for harder SiC ceramics. Grain refinement and reduced amount and size of pores resulted for superior resistance against erosion. Erosion rates of SiC ceramics were estimated by semi-empirical models with mechanical properties like hardness, fracture toughness and elastic modulus. Since none of the proposed models for erosion prediction/estimation was completely satisfactory, it advocated necessary incorporation of microstructural aspects into erosion theories. Incorporation of SiC fibers in SiC matrix resulted in improved erosion resistance at elevated temperature. The difference in erosion rate with change in impingement angle of erodent also found to be less at high temperature. In general, erosion rate of SiC composites increased with increase in temperature up to 800°C and attributed to brittle fracture and stripping of aggregates. Further increase in temperature up to 1400°C resulted in the formation of silica based glassy film and decreased erosion.

Solid particle erosion for SiC ceramics is reported to occur by elastic-plastic deformation based microfracture followed

by subsurface crack formation beneath the plastic zone. Size and type of cracks are found to be dependent on erodent shape and size, mass flow rate, velocity and angle of impingement of erodent, temperature, hardness toughness and microstructure of target materials. In case of low velocities of erodent, dependent on shape of the erodent, two types of cracks were generally reported. On impact of blunt erodent, generation of cone cracks are observed whereas lateral and/or median cracks are generated beneath the plastically deformed surface against sharp erodent. For high velocities impact of erodent, material removal from target surface is removed by plowing, severe cracking and chipping. Erosion wear of SiC ceramics was dependent on the characteristics of second phase such as geometry, durability, energy dissipation etc. During erosion, SiC fiber reinforced SiC composites initially detached out of matrix which further left pores and then resulted in easy removal of SiC ceramics. Shape, size and size distribution of erodent particles influence erosion of SiC ceramic composites.

The future scope of tribological characteristics of SiC ceramics and composites can be realized. The incorporation of microstructural parameters in contemporary wear model/theories is essential to estimate material removal in sliding or erosion conditions. Most of the studies reported high temperature wear behavior using mechanical and microstructural characteristics of SiC ceramics estimated at ambient temperatures, whereas the effect of mechanical and microstructural features at elevated temperatures is to be considered for estimating true potential. Furthermore, SiC ceramics and composites have not been considerably studied for understanding their behavior in low temperature wear applications. A future study on the effect of mechanical or microstructural characteristics at low temperature would explore the potential of SiC ceramics and composites for their use in space or marine applications. Also, the physics of degradation in different wear conditions has to be thoroughly understood. Considering expected improvement of properties in nano composites, the wear behavior of SiC nano ceramics or nano composites is also to be investigated. Studies can also be extended to understand the performance SiC based cermets for their attractive combination of hardness and fracture toughness. Material removal mechanisms of SiC ceramics in other types of wear like cavitation wear, fretting wear, slurry erosion, corrosion wear etc. are to be thoroughly understood. While most of the reported studies are focused on understanding the behavior in aerospace, automotive, chemical, manufacturing, defense applications, exploring tribological potential of SiC composites for their use in other important fields like nuclear, biomedical, electronics etc. is also a prospective future direction.

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REFERENCES

1. G. Roewer, U. Herzog, K. Trommer, E. Muller, and S. Fruhauf, "Silicon Carbide—A Survey of Synthetic Approaches, Properties and Applications," *Struct. Bond.*, **101** 59-135 (2002).
2. B. V. M. Kumar, Y.-W. Kim, D. S. Lim, and W. S. Seo, "Influence of Small Amount of Sintering Additives on Unlubricated Sliding Wear Properties of SiC Ceramics," *Ceram. Int.*, **37** [8] 3599-608 (2011).
3. N. P. Padture, "In Situ-Toughened Silicon Carbide," *J. Am. Ceram. Soc.*, **77** 519-23 (1994).
4. P. T. B. Shaffer, "A Review of the Structure of Silicon Carbide," *Acta Cryst., B*, **25** 477-88 (1969).
5. H. P. Iwata, U. Lindefelt, S. Oberg, and P.R. Briddon, "Stacking Faults in Silicon Carbides," *Phys. B*, **340-342** 165-70 (2003).
6. Y. M. Tairov and V. F. Tsvetkov, "Investigation of Growth Processes of Ingots of Silicon Carbide Single Crystals," *J. Cryst. Growth*, **43** [2] 209-12 (1978).
7. A. J. Van Bommel, J. E. Crombeen, and A. V. Tooren, "Leed and Auger Electron Observations of the SiC (0001) Surface," *Surf. Sci.*, **48** 463-72 (1975).
8. A. Taylor and D. S. Laidler, "The Formation and Crystal Structure of Silicon Carbide," *Br. J. Appl. Phys.*, **1** [7] 174-81 (1950).
9. W. Wesch, "Silicon Carbide: Synthesis and Processing," *Nucl. Instrum. Meth. B*, **116** 305-21 (1996).
10. V. A. Izhevskiy, L. A. Genova, J. C. Bressiani, and A. H. A. Bressiani, "Silicon Carbide. Structure, Properties and Processing," *Ceramica*, **46** 4-13 (2000).
11. T. Hase, "Boron Transport and Change of Lattice Parameter during Sintering of β -SiC," *J. Am. Ceram. Soc.*, **63** [5-6] 349-50 (1980).
12. K. Negita, "Effective Sintering Aids for Silicon Carbide Ceramics: Reactivities of Silicon Carbide with Various Additives," *J. Am. Ceram. Soc.*, **69** [12] C308-10 (1986).
13. D. Sciti, S. Guicciardi, and A. Bellosi, "Effect of Annealing Treatment on Microstructure and Mechanical Properties of Liquid-Phase Sintered Silicon Carbide," *J. Eur. Ceram. Soc.*, **21** 621-32 (2001).
14. D. I. Cheong, J. Kim, and S. J. Kang, "Effect of Isothermal Annealing on the Microstructure and Mechanical Properties of SiC Ceramics Hot-Pressed with Y_2O_3 and Al_2O_3 Additions," *J. Eur. Ceram. Soc.*, **22** 1321-27 (2002).
15. D. Sciti and A. Bellosi, "Effects of Additives on Densification, Microstructure and Properties of Liquid-Phase Sintered Silicon Carbide" *J. Mater. Sci.*, **35** 3849-55 (2000).
16. G. Rixecker, I. Wiedmann, A. Rosinus, and F. Aldinger, "High-Temperature Effect in the Fracture Mechanical Behaviour of Silicon Carbide Liquid-Phase Sintered with AlN - Y_2O_3 Additives," *J. Eur. Ceram. Soc.*, **21** 1013-19 (2001).
17. R. R. Lee and W. C. Wei, "Fabrication, Microstructure, and Properties of SiC- AlN Ceramic Alloys," *Ceram. Eng. Sci. Proc.*, **11** [7-8] 1094-121 (1990).
18. S. G. Lee, Y.-W. Kim, and M. Mitomo, "Relationship between Microstructure and Fracture Toughness of Toughened Silicon Carbide Ceramics," *J. Am. Ceram. Soc.*, **84** 1347-53 (2001).
19. H. J. Choi, J. G. Lee, and Y.-W. Kim, "Oxidation Behavior of Liquid-Phase Sintered Silicon Carbide with Aluminum Nitride and Rare-Earth Oxides (Re_2O_3 , where $Re = Y, Er, Yb$)," *J. Am. Ceram. Soc.*, **85** [9] 2281-86 (2002).
20. Y. Zhou, K. Hirao, Y. Yamauchi, and S. Kanzaki, "Tailoring the Mechanical Properties of Silicon Carbide Ceramics by Modification of the Intergranular Phase Chemistry and Microstructure," *J. Eur. Ceram. Soc.*, **22** 2689-96 (2002).
21. D. Chen, X. F. Zhang, and R. O. Ritchie, "Effects of Grain-Boundary Structure on the Strength, Toughness, and Cyclic-Fatigue Properties of a Monolithic Silicon Carbide," *J. Am. Ceram. Soc.*, **83** 2079-81 (2000).
22. R. Yuan, J. J. Kruzic, X. F. Zhang, L. C. De Jonghe, and R. O. Ritchie, "Ambient to High Temperature Fracture Toughness and Cyclic Fatigue Behavior in Al Containing Silicon Carbide Ceramics," *Acta Mater.*, **51** 6477-91 (2003).
23. P. F. Becher, "Microstructural Design of Toughened Ceramics," *J. Am. Ceram. Soc.*, **74** 255-269 (1991).
24. H. J. Choi, Y. -W. Kim, M. Mitomo, T. Nishimura, J. H. Lee, and D. Y. Kim, "Intergranular Glassy Phase Free SiC Ceramics Retains Strength at 1500°C," *Scripta Mater.*, **50** 1203-7 (2004).
25. Y. -W. Kim, S. H. Lee, T. Nishimura, and M. Mitomo, "Heat-Resistant Silicon Carbide with Aluminum Nitride and Scandium Oxide," *Acta Mater.*, **53** 4701-8 (2005).
26. Y.-W. Kim, J. H. Lee, and D. Y. Kim, "Effect of Sintering Additive Composition on Grain Boundary Structure in Liquid-Phase-Sintered Silicon Carbide," *Mater. Sci. Forum*, **558-559** 897-902 (2007).
27. B. V. M. Kumar, M. H. Roh, Y.-W. Kim, W. Kim, S. W. Park, and W. S. Seo, "Effect of Additive Composition on Microstructure and Mechanical Properties of SiC Ceramics Sintered with Small Amount of RE_2O_3 (RE: Sc, Lu, Y) and AlN ," *J. Mater. Sci.*, **44** 5939-5943 (2009).
28. Y.-W. Kim, Y. S. Chun, T. Nishimura, M. Mitomo, and Y. H. Lee, "High Temperature Strength of Silicon Carbide Ceramics Sintered with Rare-Earth Oxide and Aluminum Nitride," *Acta Mater.*, **55** 727-36 (2007).
29. N. P. Padture, "In Situ-Toughened Silicon Carbide," *J. Am. Ceram. Soc.*, **77** 519-23 (1994).
30. G. Rixecker, K. Biswas, A. Rosinus, S. Sharma, I. Wiedmann, and F. Aldinger, "Fracture Properties of SiC Ceramics with Oxynitride Additives," *J. Eur. Ceram. Soc.*, **22** 2669-75 (2002).
31. K. Strecker and M. J. Hoffmann, "Effect of AlN -Content on the Microstructure and Fracture Toughness of Hot-Pressed a Heat-Treated LPS-SiC Ceramics," *J. Eur. Ceram. Soc.*, **25** 801-7 (2005).
32. M. Herrmann, G. Standke, S. Hohn, G. Himpel, and T. Gestrich, "High-Temperature Corrosion of Silicon Carbide Ceramics by Coal Ashes," *Ceram. Int.*, **40** 1471-79 (2014).
33. Y. -W. Kim, K. Y. Lim, and W. S. Seo, "Microstructure

- and Thermal Conductivity of Silicon Carbide with Yttria and Scandia," *J. Am. Ceram. Soc.*, **97** 923-28 (2014).
34. K. Y. Lim, Y.-W. Kim, and K. J. Kim, "Mechanical Properties of Electrically Conductive Silicon Carbide Ceramics," *Ceram. Int.*, **40** 10577-82 (2014).
 35. S. Grasso, T. Saunders, H. Porwal, and M. Reece, "Ultra-High Temperature Spark Plasma Sintering of α -SiC," *Ceram. Int.*, **41** 225-30 (2015).
 36. S. Lafon-Placette, K. Delbe, J. Denape, and M. Ferrato, "Tribological Characterization of Silicon Carbide and Carbon Materials," *J. Eur. Ceram. Soc.*, **35** 1147-59 (2015).
 37. S. M. Hsu and M. Sheng, "Wear Prediction of Ceramics," *Wear*, **256** 867-78 (2004).
 38. D. Chen, X. F. Zhang, and R. O. Ritchie, Effects of Grain-Boundary Structure on the Strength, Toughness, and Cyclic-Fatigue Properties of a Monolithic Silicon Carbide," *J. Am. Ceram. Soc.*, **83** 2079-81 (2000).
 39. X. Dong, S. Jahanmir, and L. K. Ives, "Wear Transition Diagram for Silicon Carbide," *Tribol. Int.*, **28** 559-72 (1995).
 40. D. C. Cranmer, "Friction and Wear Properties of Monolithic Silicon-Based Ceramics," *J. Mater. Sci.*, **20** 2029-37 (1985).
 41. K. H. Z. Gahr, R. Blattner, D. H. Hwang, and K. Pohlmann, "Micro- and Macro-Tribological Properties of SiC Ceramics in Sliding Contact," *Wear*, **250** 299-310 (2001).
 42. V. S. R. Murthy, H. Kobayashi, S. Tsurekawa, N. Tamari, T. Watanabe, and K. Kato, "Influence of Humidity and Hoping Elements on the Friction and Wear of SiC in Unlubricated Sliding," *Tribol. Int.*, **37** 353-64 (2004).
 43. Y. Wang and S. M. Hsu, "Wear and Wear Transition Mechanisms of Ceramics," *Wear* **195** 112-22 (1996).
 44. X. F. Zhang, G. Y. Lee, D. Chen, R. O. Ritchie, and L. C. D. Jonghe, "Abrasive Wear Behavior of Heat-Treated ABC-Silicon Carbide," *J. Am. Ceram. Soc.*, **86** 1370-78 (2003).
 45. P. Andersson and A. Blomberg, "Instability in the Tribochemical Wear of Silicon Carbide in Unlubricated Sliding Contacts," *Wear*, **174** 1-7 (1994).
 46. O. O. Adewoye and T. F. Page, "Frictional Deformation and Fracture in Polycrystalline SiC and Si_3N_4 ," *Wear*, **70** 37-5 (1981).
 47. S. J. Cho, C. D. Um, and S. S. Kim, "Wear and Wear Transition in Silicon Carbide Ceramics During Sliding," *J. Am. Ceram. Soc.*, **79** 1247-51 (1996).
 48. O. B. Lopez, A. L. Ortiz, F. Guiberteau, and N. P. Padture, "Sliding-Wear-Resistant Liquid-Phase-Sintered SiC Processed Using α -SiC Starting Powders," *J. Am. Ceram. Soc.*, **90** [2] 541-45 (2007).
 49. O. B. Lopez, A. L. Ortiz, F. Guiberteau, and N. P. Padture, "Microstructural Design of Sliding-Wear-Resistant Liquid-Phase-Sintered SiC: An Overview," *J. Eur. Ceram. Soc.*, **27** 3351-57 (2007).
 50. O. B. Lopez, A. L. Ortiz, F. Guiberteau, and N. P. Padture, "Effect of Microstructure on Sliding-Wear Properties of Liquid-Phase-Sintered α -SiC," *J. Am. Ceram. Soc.*, **88** [8] 2159-2163 (2005).
 51. O. B. Lopez, A. L. Ortiz, F. Guiberteau, and N. P. Padture, "Improved Sliding-Wear Resistance in In-situ Toughened Silicon Carbide," *J. Am. Ceram. Soc.*, **88** [12] 3531-34 (2005).
 52. O. B. Lopez, A. L. Ortiz, F. Guiberteau, and N. P. Padture, "Effect of the Nature of the Intergranular Phase on Sliding-Wear Resistance of Liquid-Phase-Sintered α -SiC," *Scripta Mater.*, **57** [6] 505-8 (2007).
 53. E. Ciudad, O. B. Lopez, F. Rodriguez-Rojas, A. L. Ortiz, and F. Guiberteau, "Effect of Intergranular Phase Chemistry on the Sliding-Wear Resistance of Pressureless Liquid-Phase-Sintered α -SiC," *J. Eur. Ceram. Soc.*, **32** [2] 511-16 (2012).
 54. A. L. Ortiz, O. B. Lopez, M. Z. Quadir, and F. Guiberteau, "A Route for the Pressureless Liquid-Phase Sintering of SiC with Low Additive Content for Improved Sliding-Wear Resistance," *J. Eur. Ceram. Soc.*, **32** [4] 965-73 (2012).
 55. R. S. Gates and S. M. Hsu, "Tribochemistry between Water and Si_3N_4 and SiC: Induction Time Analysis," *Tribol. Lett.*, **17** [3] 399-407 (2004).
 56. S. M. Hsu and M. C. Shen, "Ceramic Wear Maps," *Wear*, **200** [1-2] 154-75 (1996).
 57. K. Kato and K. Adachi, "Wear of Advanced Ceramics," *Wear*, **253** [11-12] 1097-104 (2002).
 58. L. Jordi, C. Iliev, and T. E. Fischer, "Lubrication of Silicon Nitride and Silicon Carbide by Water: Running in, Wear and Operation of Sliding Bearings," *Tribol. Lett.*, **17** [3] 367-76 (2004).
 59. A. Udayakumar, M. Balasubramanian, H. B. Gopala, P. Sampathkumaran, S. Seetharamu, R. Babu, D. Sathiyamoorthy, and G. R. Reddy, "Influence of the Type of Interface on the Tribological Characteristics of ICVI Generated SiC/SiC Composites," *Wear*, **271** 859-65 (2011).
 60. U. Dulias and K. H. Z. Gahr, "Investigation of Al_2O_3 - and SSiC-Ceramic Under Lubricated, Reciprocating Sliding Contact and Cavitation Erosion," *Mater. Sci. Eng. Technol.*, **36** [3-4] 140-47 (2005).
 61. S. K. Sharma, B. V. M. Kumar, K. Y. Lim, Y.-W. Kim, and S. K. Nath, "Erosion Behavior of SiC-WC Composites," *Ceram. Int.*, **40** 6829-39 (2014).
 62. J. J. Kim and S.K. Park, "Solid Particle Erosion of SiC and SiC-TiB₂ Composite Hot-Pressed with Y_2O_3 ," *Wear*, **222** [2] 114-19 (1998).
 63. X. Li, H. Ding, Z. Huang, M. Fang, B. Liu, Y. Liu, X. Wu, and S. Chen, "Solid Particle Erosion-Wear Behavior of SiC-Si₃N₄ Composite Ceramic at Elevated Temperature," *Ceram. Int.*, **40** 16201-7 (2014).
 64. B. Wang and A.V. Levy, "Erosion Behavior of SiC Fiber-SiC Matrix Composites," *Wear*, **138** 125-36 (1990).
 65. D. F. Wang, J. H. She, and Z. Y. Ma, "Effect of Microstructure on Erosive Wear Behavior of SiC Ceramics," *Wear*, **180** 35-41 (1995).
 66. J. L. Routbort and A. P. L. Turner, "The Erosion Rate of Reaction-Bonded SiC Containing Various Amounts of Free Silicon," *Wear*, **84** 381-85 (1983).
 67. J. L. Routbort, R. O. Scattergood, and A. P. L. Turner, "The Erosion of Reaction-Bonded SiC," *Wear*, **59** 363-75 (1980).

68. B. Basu and M. Kalin, "Tribology of Ceramics and Composites," pp. 70-100 in *Wear Mechanisms*, John Wiley & Sons, Inc., New Jersey, 2011.
69. S. J. Cho, B. J. Hockey, B. R. Lawn, and S. J. Bennison, "Grain-Size and R-Curve Effects in the Abrasive Wear of Alumina," *J. Am. Ceram. Soc.*, **72** [7] 1249-52 (1989).
70. S. J. Cho, C. D. Um, and S. S. Kim, "Wear and Wear Transition Mechanism in Silicon Carbide during Sliding," *J. Am. Ceram. Soc.*, **78** [4] 1076-78 (1995).
71. J. D. O. Barceinas-Sanchez and W. M. Rainforth, "On the Role of Plastic Deformation during the Mild Wear of Alumina," *Acta Mater.*, **46** [18] 6475-83 (1998).
72. V. S. R. Murthy, H. Kobayashi, N. Tamari, S. Tsurekawa, T. Watanabe, and K. Kato, "Effect of Doping Elements on the Friction and Wear Properties of SiC in Unlubricated Sliding Condition," *Wear*, **257** 89-96 (2004).
73. R. K. Iyer, *The Chemistry of Silica*; pp. 40-60, Wiley Inc., New York, 1979.
74. E. Ciudad, O. B. Lopez, A. L. Ortiz, and F. Guiberteau, "Microstructural Effects on the Sliding-Wear Resistance of Pressureless Liquid-Phase-Sintered SiC Under Diesel Fuel," *J. Eur. Ceram. Soc.*, **33** [4] 879-85 (2013).
75. Y. Zhou, K. Hirao, Y. Yamauchi, and S. Kanzaki, "Tribological Properties of Silicon Carbide and Silicon Carbide-Graphite Composite Ceramics in Sliding Contact," *J. Am. Ceram. Soc.*, **86** [6] 991-1002 (2003).
76. S. K. Sharma, B. V. M. Kumar, and Y.-W. Kim, "Effect of WC Addition on Sliding Wear Behavior of SiC Ceramics," *Ceram. Int.*, **41** 3427-37 (2015).
77. X. Guo, H. Yang, L. Zhang, and X. Zhu, "Sintering Behavior, Microstructure and Mechanical Properties of Silicon Carbide Ceramics Containing Different Nano-TiN Additive," *Ceram. Int.*, **36** 161-65 (2010).
78. H. J. Yeom, Y.-W. Kim, and K. J. Kim, "Electrical, Thermal and Mechanical Properties of Silicon Carbide-Silicon Nitride Composites Sintered with Yttria and Scandia," *J. Eur. Ceram. Soc.*, **35** 77-86 (2015).
79. J. Llorente, B. R. Manso, P. Miranzo, and M. Belmonte, "Tribological Performance under Dry Sliding Conditions of Graphene/Silicon Carbide Composites," *J. Eur. Ceram. Soc.*, **36** 429-35 (2016).
80. X. Yang, X. Liu, L. Wang, H. Zhang, X. Yao, and Z. Huang, "R-Curve Analysis of Solid-Phase-Sintered and Liquid-Phase-Sintered Silicon Carbide Ceramics by Indentation Fracture and Indentation-Strength-in-Bending Methods," *Ceram. Int.*, **42** 4011-18 (2016).
81. D. B. Marshall, B. R. Lawn, and A. G. Evans, "Elastic/Plastic Indentation Damage in Ceramics: the Lateral Crack System," *J. Am. Ceram. Soc.*, **65** [11] 561-66 (1982).
82. B. R. Lawn, N. P. Padture, F. Guiberteau, and H. Cai, "A Model for Microcrack Initiation and Propagation Beneath Hertzian Contacts in Polycrystalline Ceramics," *Acta Metall. Mater.*, **42** [5] 1683-93 (1994).
83. A. Tewari, B. Basu, and R. K. Bordia, "Model for Fretting Wear of Brittle Ceramics," *Acta Mater.*, **57** 2080-87 (2009).
84. S. Gupta, S. K. Sharma, B. V. M. Kumar, and Y.-W. Kim, "Tribological Characteristics of SiC Ceramics Sintered with a Small Amount of Yttria," *Ceram. Int.*, **41** 14780-89 (2015).
85. J. F. Li, J. Q. Huang, S. H. Tan, Z. M. Cheng, and C. X. Ding, "Tribological Properties of Silicon Carbide under Water-Lubricated Sliding," *Wear*, **218** 167-71 (1998).
86. A. Kovalcikova, P. Kurek, J. Balko, J. Dusza, P. Sajgalik, and M. Mihalikova, "Effect of the Counterpart Material on Wear Characteristics of Silicon Carbide Ceramics," *Int. J. Refract. Met Hard Mater.*, **44** 12-8 (2014).
87. G. Amirthan and M. Balasubramanian, "Reciprocating Sliding Wear Studies on Si/SiC Ceramic Composites," *Wear*, **271** 1039-49 (2011).
88. V. M. Candelario, O. B. Lopez, F. Guiberteau, R. Morenob, and A. L. Ortiz, "Sliding- Wear Resistance of Liquid-Phase-Sintered SiC Containing Graphite Nanodispersoids," *J. Eur. Ceram. Soc.*, **34** 2597-602 (2014).
89. J. Takadom, Z. Zsiga, and C. Roques-Carmes, "Wear Mechanism of Silicon Carbide: New Observations," *Wear*, **174** 239-42 (1994).
90. L. Micele, G. Palombarini, S. Guicciardi, and L. Silvestroni, "Tribological Behaviour and Wear Resistance of a SiC-MoSi₂ Composite Dry Sliding against Al₂O₃," *Wear*, **269** 368-75 (2010).
91. R. Wasche, D. Klaffke, and T. Troczynski, "Tribological Performance of SiC and TiB₂ against SiC and Al₂O₃ at Low Sliding Speeds," *Wear*, **256** 695-704 (2004).
92. K. Kato, "Tribology of Ceramics," *Wear*, **136** 117-33 (1990).
93. S. Hotta and Y. Mizutani, "Effect of Water and Organic Solvents on the Friction of SiC," pp. 1-4 in *Proceedings, 32nd JSLE, Japanese Society of Lubrication Engineers Conference*, Tokyo, Japan (1988).
94. X. F. Chen, X. G. Xu, X. B. Hu, J. Li, S. Z. Jiang, L. N. Ning, Y. M. Wang, and M. H. Jiang, "Anisotropy of Chemical Mechanical Polishing in Silicon Carbide Substrates," *Mater. Sci. Eng. B*, **142** 28-30 (2007).
95. H. Y. Liu and M. E. Fine, "Modeling of Grain-Size Dependent Microfracture-Controlled Sliding Wear in Polycrystalline Alumina," *J. Am. Ceram. Soc.*, **76** [9] 2393-96 (1993).
96. S. J. Cho, H. Moon, B. J. Hockey, and S. M. Hsu, "The Transition from Mild to Severe Wear in Alumina During Sliding," *Acta Metall.*, **40** [1] 185-92 (1992).
97. X. Dong and S. Jahanmir, "Wear Transition Diagram for Silicon Nitride," *Wear*, **165** 169-80 (1993).
98. W. M. Rainforth, "The Wear Behavior of Oxide Ceramics-A Review," *J. Mater. Sci.*, **39** [2] 6705-21 (2004).
99. X. Wang, N. P. Padture, H. Tanaka, and A. L. Ortiz, "Wear-Resistant Ultra-Fine-Grained Ceramics," *Acta Mater.*, **53** [2] 271-77 (2005).
100. B. Frisch, W. R. Thiele, R. Drumm, and B. Muennich, "On the Oxidation Mechanism of Silicon Carbide in the 300°C to 1300°C Temperature Range," *Ceram. Forum Int.*, **65** [8-9] 277-84 (1988).
101. S. C. Singhal, "Thermodynamical Analysis of the High-Temperature Stability of Silicon Nitride and Silicon Carbide," *Ceram. Int.*, **2** [3] 123-30 (1976).
102. P. C. Kong and E. Pfender, "Formation of Ultrafine β -Silicon Carbide Powders in an Argon Thermal Plasma Jet," *Langmuir*, **3** [2] 259-65 (1987).
103. Z. Zhu, V. Muratov, and T. E. Fischer, "Tribochemical

- Polishing of Silicon Carbide in Oxidant Solution," *Wear*, **225-229** 848-56 (1999).
104. L. C. Erickson, A. Blomberg, S. Hogmark, and J. Bratthall, "Tribological Characterization of Alumina and Silicon Carbide under Lubricated Sliding," *Tribology Int.*, **26** [2] 83-92 (1993).
 105. H. Tomizawa and T. E. Fischer, "Friction and Wear of Silicon Nitride and Silicon Carbide in Water: Hydrodynamic Lubrication at Low Sliding Speed Obtained by Tribochemical Wear," *ASLE Trans.*, **30** [1] 41-6 (1987).
 106. J. K. Lancaster, "A Review of the Influence of Environmental Humidity and Water on Friction, Lubrication and Wear," *Tribology Int.*, **23** [6] 371-89 (1990).
 107. S. Sasaki, "The Effects of Water on Friction and Wear of Ceramics," *J. Jpn. Soc. Lubr. Eng.*, **10** 10-26 (1988).
 108. R. Wasche and D. Klaffke, "Ceramic Particulate Composites in the System SiC-TiC-TiB₂ Sliding against SiC and Al₂O₃ Under Water," *Tribology Int.*, **32** 197-206 (1999).
 109. E. Medvedovski, "Ballistic Performance of Armour Ceramics: Part 1. Influence of Design and Structure," *Ceram. Int.*, **36** 2103-15 (2010).
 110. A. G. Evans and T. R. Wilshaw, "Dynamic Solid Particle Damage in Brittle Materials, An Appraisal," *J. Mater. Sci.*, **12** 97-116 (1977).
 111. J. G. Chacon-Nava, F. H. Stott, S. D. de la Torre, and A. Martinez-Villafane, "Erosion of Al₂O₃ and SiC at Low Impact Velocities," *Mater. Lett.*, **55** 269-73 (2002).
 112. J. L. Routbort and R. O. Scattergood, "Anomalous Solid-Particle Erosion Rate of Hot Pressed Silicon Carbide," *J. Am. Ceram. Soc.*, **63** 593-95 (1980).
 113. S. M. Wiederhorn and B. J. Hockey, "Effect of Material Parameters on the Erosion Resistance of Brittle Materials," *J. Mater. Sci.*, **18** 766-80 (1983).
 114. J. L. Routbort and H. Matzke, "On the Correlation Between Erosion and Fracture Parameters in SiC Ceramics," *J. Mater. Sci.*, **18** 1491-96 (1983).
 115. D. Jianxin, L. Lili, and D. Mingwei, "Erosion Wear Behaviours of SiC/(W,Ti)C Laminated Ceramic Nozzles in Dry Sand Blasting Processes," *Mater. Sci. Eng. A*, **444** 120-29 (2007).
 116. A. F. Colclough and J. A. Yeomans, "Hard Particle Erosion of Silicon Carbide and Silicon Carbide-Titanium Diboride from Room Temperature to 1000°C," *Wear*, **209** 229-36 (1997).
 117. I. M. Hutchings, "Tribology: Friction and Wear of Engineering Materials; pp. 352, Edward Arnold, London, 1992.
 118. J. F. Bell, and P. S. Rogers, "Laboratory Scale Erosion Testing of a Wear Resistant Glass-Ceramic," *Mater. Sci. Tech.*, **3** 807-13 (1987).
 119. P. Gautier and K. Kato, "Wear Mechanisms of Silicon Nitride Partially Stabilized Zirconia and Alumina in Unlubricated Sliding against Steel," *Wear*, **162-164** 305-13 (1993).
 120. D. K. Shetty, I. G. Wright, and A. H. Clauer, "Coal Slurry Erosion of Reaction-Bonded SiC," *Wear*, **79** 275-79 (1982).
 121. B. Q. Wang and A. V. Levy, "Erosion Behavior of SiC Fiber-SiC Matrix Composites," *Wear*, **138** 125-36 (1990).
 122. I. Finnie, J. Wolak, and Y. H. Kabil, "Erosion of Metals by Solid Particles," *J. Mater.*, **2** [3] 682-700 (1967).
 123. G. Amirthana, A. Udaya Kumar, V.V. Bhanu Prasad, and M. Balasubramanian, "Solid Particle Erosion Studies on Biomorph Si/SiC Ceramic Composites," *Wear*, **268** 145-52 (2010).
 124. A. R. de A. Lopez, J. M. Fernandez, F. M. V. Fera, T. S. Orlova, K. C. Goretta, F. G. Mora, Nan Chen, and J. L. Routbort, "Erosion and Strength Degradation of Biomorph SiC," *J. Eur. Ceram. Soc.*, **24** 861-70 (2004).
 125. B. J. Hockey, S. M. Wiederhorn, and H. Johnson, "Erosion of Brittle Materials by Solid Particle Impact," pp. 379-402 in *Fracture Mechanics of Ceramics Vol. 3*, Ed. by R. C. Bradt, D. P. H. Hasselman, and F. F. Lange, Plenum Press, New York, 1978.
 126. B. J. Hockey and B. R. Lawn, "Electron Microscopy of Microcracking about Indentations in Aluminium Oxide and Silicon Carbide," *J. Mater. Sci.*, **10** [8] 1275-84 (1975).
 127. S. M. Wiederhorn and B. J. Hockey, "Effect of Material Parameters on the Erosion Resistance of Brittle Materials," *J. Mater. Sci.*, **18** [3] 766-80 (1983).
 128. B. R. Lawn, B. J. Hockey, and S. M. Wiederhorn, "Atomically Sharp Cracks in Brittle Solids: An Electron Microscopy Study," *J. Mater. Sci.*, **15** 1207-23 (1980).
 129. D. C. Evans and J. K. Lancaster, "The Wear of Polymers," pp. 85-139 in *Treatise on Materials Science and Technology: Wear*, Vol. 13, Academic Press, New York and London, 1979.
 130. D. B. Marshall, B. R. Lawn, and A. G. Evans, "Elastic/Plastic Indentation Damage in Ceramics: The Lateral Crack System," *J. Am. Ceram. Soc.*, **65** [11] 561-66 (1982).
 131. A. G. Evans, M. E. Gulden, and M. Rosenblatt, "Impact Damage in Brittle Materials in the Elastic-Plastic Response Regime," *Proc. of the Royal Society of London*, **361** 343-65 (1978).
 132. D. Wang and Z. Mao, "Microscopic Observations of Wear of Heat-Resistant Ceramics," *Wear*, **167** [1] 87-9 (1993).
 133. M. M. Chaudhri and S. M. Walley, "Damage to Glass Surfaces by the Impact of Small Glass and Steel Spheres," *Phil. Mag. A*, **37** [2] 153-65 (1978).
 134. C. G. Knight, M. V. Swain, and M. M. Chaudhri, "Impact of Small Steel Spheres on Glass Surfaces," *J. Mater. Sci.*, **12** [8] 1573-86 (1977).
 135. B. R. Lawn and D. B. Marshall, "Indentation Fracture and Strength Degradation in Ceramics," pp. 205-29 in *Fracture Mechanics of Ceramics*, Vol. 3, Ed. by R. C. Bradt, D. P. H. Hasselman, and F. F. Lange, Plenum Press, New York, 1978.
 136. M. M. Chaudhri, and P. A. Brophy, "Single Particle Impact Damage of Fused Silica," *J. Mater. Sci.*, **15** [2] 345-52 (1980).
 137. M. S. Suh, T. Hinoki, and A. Kohyama, "Erosive Wear Mechanism of New SiC/SiC Composites by Solid Particles," *Tribol. Lett.*, **41** 503-13 (2011).
 138. S. Gochnour, J. D. Bright, D. K. Shett, "Solid Particle Erosion of SiC-Al₂O₃ Ceramics," *J. Mater. Sci.*, **25** 3229-35 (1990).
 139. S. Wada and N. Watanabe, "Solid Particle Erosion of Brittle Materials. III. The Interaction with Material

- Properties of Target and That of Impingement Particle on Erosive Wear Mechanism,” *J. Ceram. Soc. Jpn.*, **95** [6] 573-78 (1987).
140. J. J. Kim and S. K. Park, “Solid-Particle Erosion of Hot-Pressed Silicon Carbide and SiC-TiB₂ Composite,” *J. Mater. Sci. Lett.*, **16** 821-23 (1997).
141. Z. H. Huang, J. L. Sun, J. X. Wang, and Y. R. Hong, “ α -Sialon-Al₂O₃-SiC Composite Refractories,” *Key Eng. Mater.*, **224** 275-80 (2012).