High-Temperature Structural Ceramics

R. Nathan Katz

Summary. The unique properties of ceramics based on silicon carbide and silicon nitride make them prime candidates for use in advanced energy conversion systems. These compounds are the bases for broad families of engineering materials, whose properties are reviewed. The relationships between processing, microstructure, and properties are discussed. A review and assessment of recent progress in the use of these materials in high-temperature engineering systems, and vehicular engines in particular, is presented.

Accelerated research activity over the past decade on the "new" high-temperature structural ceramics may lead to one of those rare instances in the history of technology when the appropriate materials and engineering technology come together at the right time to meet urgent societal needs. Availability of energy, availability of critical materials, and maintaining a livable environment are, perhaps, the three most urgent problems facing the industrialized world. It is widely recognized that the new high-temperature structural ceramics based on silicon carbide, SiC, and silicon nitride, Si_3N_4, offer the possibility of helping to resolve the first two of these problems (*1-4*), with no negative effects [and maybe some slight positive ones (*5*)] on the third.

The silicon carbide- and silicon nitride-based ceramics have unique combinations of properties that make them very attractive to designers of vehicular engines, energy conversion systems, and industrial heat exchangers. In particular, they provide high strength at high temperature, good thermal stress resistance, and excellent oxidation, corrosion, and erosion resistance. This combination of properties can be used to increase the operating temperature, or reduce the heat lost to cooling, in gas turbine, diesel, and Stirling engines and in industrial heat exchangers, thereby yielding more power per unit of fuel. The low density of these materials (~ 40 percent that of high-temperature superalloys) may offer components of lower weight and inertia, which would translate into improved performance for conventional automotive engines. The universal abundance and inherent low cost of the elements silicon, carbon, and nitrogen contrast sharply with the increasingly short supply and increasing costs of critical metals for high-temperature alloys such as chromium, nickel, cobalt, and tungsten. Table 1 gives some perspective on the fuel savings that the use of these ceramics could produce in several applications (*6-10*). Recent estimates (*10, 11*) indicate that fuel savings of 1/2 billion barrels of oil per year, amounting to $17.5 billion at current prices, can be realized in the areas of highway vehicles and industrial heat exchangers if this technology is fully exploited. Thus there is considerable incentive to apply these ceramics in high-temperature energy conversion technology.

So far, I have made silicon carbide and silicon nitride ceramics sound like panaceas. If this class of materials can do all of the above, why are they not used in today's engines, other energy conversion devices, and industrial heat exchangers? The answer lies in the brittleness of ceramic materials—but this is a bit too simplistic an answer. All engineering design represents a trade-off in which (we hope) the best compromise in the use of materials and design emerges. The practice in engineering design to date has been to take advantage of the fact that metals are forgiving of contact stresses and stress concentrations (that is, they relieve local overstresses by local yielding). This simplifies the design process. The trade-off is that metals soften or melt and must be cooled for high-temperature use. Management of the cooling fluid adds design complexity and weight and drains power. Up to now, the compromises in design that have been required to use metals have had a net payoff. However, in either economic or actual physical terms, the end of this line of development is in sight. Further development of heat engines, in particular, requires higher temperatures with reduced cooling and at acceptable costs. Thus if a combination of improved ceramic materials and the ability of designers to create compliant designs to accommodate the noncompliant nature (brittleness) of ceramics exists, then the materials-design trade-offs will favor ceramics. Partly because of the brittleness problem, there is very little design experience with such structural applications of ceramic materials. This lack of experience has itself served to restrict the use of ceramics in these applications.

Designers have been intrigued by the possibility of applying ceramics to turbines and diesels for more than 40 years. However, it was not until about 10 years ago that the two prerequisites were present: thermal shock-resistant structural ceramics, such as SiC and Si_3N_4, developed to the point where they could begin to be considered as engineering materials, and, equally important, sufficiently developed computer capability to handle the complex job of detailing stresses with the degree of refinement required in brittle materials design. The fact that the materials and design technology were advancing at the same time that the need for this technology was being widely perceived led to a large number of technology demonstration programs in the mid-1970's (*2, 3*). I will return to some of these programs later in this article in order to assess where this technology stands.

The SiC and Si_3N_4 ceramics that are being developed for engine and heat exchanger use are actually families of ceramic materials, in much the same way as steels and brasses are alloy families. In materials families, as in all families, certain characteristics are shared. In the case of the carbides and nitrides of silicon the shared characteristics include strongly covalent bonding, dissociation rather than melting at high temperatures (> 1900°C), low coefficients of thermal expansion (which lead to the excellent thermal shock behavior of these materials), and resistance to oxidation provided by a spontaneously forming surface oxide film. These similarities aside, each individual member of these families

---

The author is chief of the Ceramics Research Division, U.S. Army Materials and Mechanics Research Center, Watertown, Massachusetts 02172.

0036-8075/80/0523-0841$01.75/0

Table 1. Examples of projected energy savings from use of ceramic technology.

Technology	Systems configuration	Reduction in fuel use (%)	Reference
Gas turbines			
Automotive	~ 150 hp, regenerated, single-shaft engine, 1370°C turbine inlet temperature	27	(*6*)
Truck	~ 350 hp, regenerated, two-shaft engine, 1240°C turbine inlet temperature	17	(*7*)
Industrial/ship	~ 1000 hp, simple-cycle, three-stage engine, 1370°C turbine inlet temperature	10	(*8*)
Diesel engines			
Truck	~ 500 hp, adiabatic-turbocompound, 1210°C maximum component temperature	22	(*9*)
Industrial heat recovery			
Recuperators	Silicon carbide recuperated slot forging furnace operating at ~1300°C	42	(*10*)

of ceramics is a unique material made by a unique process. For example, within the Si_3N_4 family we have hot-pressed, slip-cast sintered, injection-molded sintered, and injection-molded reaction-bonded materials. These materials have very different properties as a consequence of their processing. Also, the ease and cost of producing useful engineering shapes differ widely from one processing method to another. In ceramic materials, unlike metals, one cannot readily manipulate the microstructure, and hence the properties, after the primary processing of the material. Therefore, processing science and technology is central to the development of improved ceramics. The progress during the past decade on the carbides and nitrides of silicon has largely been the result of increased understanding of processing and the relationship between processing, microstructure, and properties. Therefore, to have a perspective for evaluating the progress in the development of these materials, it is important to briefly review the processing of ceramics in general.

Ceramic Processing

Virtually all high-temperature engineering ceramics are formed by the consolidation of powders. This is the case for the silicon-based structural ceramics under consideration here. The application of heat with or without pressure will, in general, cause consolidation of a ceramic powder, resulting in a strongly bonded polycrystalline body. This is known as sintering (or, if pressure is involved, hot-pressing).

The driving force for sintering is the reduction of the surface free energy of the powder aggregation. In general, sintering will occur when the surface free energy of a grain boundary, γ_{GB}, is less than twice the surface free energy of a particle surface in contact with its vapor environment (air, nitrogen, argon, and so on), γ_{SV}. Thus when the individual particles consolidate into a densified and bonded ceramic, there is a net reduction in the free energy of the system. There is also a volume shrinkage (~ 50 percent) as the voids between the particles are eliminated. If heat (or heat and pressure) is maintained, the system will continue to lower its free energy by the process known as grain growth, which will continue until a metastable grain boundary array is obtained. Often, sintering, densification, and grain growth are accelerated by the use of additives, which may or may not produce a separate grain boundary phase. Even in relatively pure single-phase ceramics, grain boundaries tend to be favored sites for segregation of impurities, porosity, and entrapped gases. Thus grain boundary composition and structure and processes occurring at grain boundaries can dictate the properties of the ceramic piece.

Covalent compounds, especially those which dissociate rather than melt at normal pressures, such as SiC and Si_3N_4, are extremely difficult to sinter to full density even with the use of additives. At temperatures low enough to avoid dissociation atomic mobility to too low for densification to occur, and at temperatures where appreciable atomic mobility is possible the compounds start to decompose. Because both silicon carbide and silicon nitride are as hard as aluminum oxide (a standard abrasive and tool bit material) they are very difficult to machine. Fabrication methods that can reduce or eliminate machining are therefore essential to significantly increased utilization of these ceramics as engineering materials. With these basic considerations in mind, we will now take a look at various members of the silicon-based families of structural ceramics.

Silicon Carbide Ceramics

Silicon carbide can be produced with either a cubic, β, or a hexagonal, α, crystal structure. The silicon carbide–based ceramics include hot-pressed, sintered, reaction-sintered, chemically vapor-deposited (CVD), and fiber-silicon composites. Typical properties of each class of SiC material are presented in Table 2.

Hot-pressed silicon carbide can be fabricated to essentially full density and high strength by using additions of boron and carbon or of aluminum oxide to either α- or β-SiC starting powder (*12, 13*). Hot-pressing is typically accomplished at temperatures of ~ 1900° to 2000°C with pressures of ~ 35 megapascals (5000 pounds per square inch). The product is surpassed in strength only by hot-pressed silicon nitrides at low to moderate temperatures, and has a higher strength than any of the silicon nitride-based ceramics in the range 1300° to 1400°C (see Table 2). Given this outstanding behavior, one would think that considerable effort would have been expended on research to improve hot-pressed SiC during the past few years. This has not been the case for several reasons. Hot-pressed SiC has the major drawback of requiring expensive diamond machining to form shaped components. This has tended to focus industrial development on developing a sinterable SiC, which can be shaped into components in the "green" (unfired, soft and relatively easily shaped) state. Further, early failures of hot-pressed SiC turbine vanes in a test where hot-pressed Si_3N_4 vanes survived (*14*) were taken as an indication that hot-pressed SiC has a lower thermal shock resistance than hot-pressed Si_3N_4, and accordingly development emphasis shifted to the nitride. In addition, the lack of general availability of sintering-grade powders (powders with the correct particle size distribution and chemistry) and the general lack of facilities for hot-pressing at > 1900°C have served to restrict basic research on this form of SiC.

The recent successful development of sintered SiC is a major accomplishment in ceramic science and technology. Until 1973 it was widely believed that SiC could not be conventionally sintered to full density because its highly covalent bond precluded the degree of volume or grain boundary diffusion required for densification. However, in that year Prochazka at General Electric (*15*) demonstrated that submicrometer β-SiC powder, with a low oxygen content (< 0.2 percent oxygen) and small additions of boron and carbon (about 0.5 and 1 percent, respectively), could be sintered to nearly full density without applying pressure at temperatures between 1950° and 2100°C under an inert gas or in a vacuum. The ultrafine powder provides a high thermodynamic driving force for densification as well as short diffusion distances, the boron appears to accelerate both volume and grain boundary diffusion, and the carbon removes the SiO_2 layer from the SiC powders. It is now widely thought that these three factors—ultrafine grain size, an additive to promote volume diffusion, and careful attention to the chemistry at the powder surface—are the key to producing other sintered, single-phase, covalent compounds such as Si_3N_4. Prochazka also demonstrated that sintered β-SiC could be formed into useful shapes by slip casting, die pressing, and extrusion. Nonuniform distribution of the carbon additive and exaggerated grain growth of α-SiC were found to interfere with densification. If one could use α starting powder rather than β, perhaps this problem could be overcome. In addition, there would be the advantage of using a less expensive, industrially available material. Coppola and McMurty (*16*) succeeded in developing such a sintered α-SiC, first reported in 1976. This material has so much promise that the Carborundum Company recently dedicated a multimillion-dollar facility to its commercialization, primarily in the area of automotive engine components. Supporting this optimism, sintered α-SiC has been successfully demonstrated in several piston engine applications.

Reaction-sintered SiC's cover a wide range of compositions and manufacturing processes. Although manufacturers use their own proprietary processes, in general a plastic body is formed of SiC powder, graphite, and a plasticizer. In some variants of the process, SiC powder plus a char-forming binder are used. The plastic body is pressed, extruded, injection-molded, or otherwise formed into a green body. The plasticizers are burned off or converted to a porous char by pyrolysis. Silicon metal as a liquid or vapor is infiltrated into the body and reacts with the graphite powder or char to form SiC in situ, which reaction-sinters the components. Excess silicon (typically 2 to 12 percent) is usually left to fill any voids, thus yielding a nonporous body. Such materials exhibit quite reasonable strengths to the melting point of silicon (~ 1400°C) or beyond, depending on the amount of free silicon retained. The presence of the free silicon is a problem in applications where temperatures above ~ 1300°C are likely to be encountered. The major advantage of these materials is that they maintain the geometry of the green preform after conversion to SiC. Thus, little machining is required and component cost is relatively low. A variety of successful experimental gas turbine components, such as combustors or stators, have been made by this materials-process route.

Table 2. Typical properties of silicon carbide and silicon nitride ceramics.

Type	Bend strength (four-point) (MPa)			Young's modulus (GPa)	Coefficient of thermal expansion, α (10^{-6} ° C^{-1})	Thermal conductivity, K (W m^{-1} ° C^{-1})
	RT*	1000°C	1375°C			
			Silicon carbides			
Hot-pressed (Al_2O_3 additive)	655	585	520	449	4.5	85-35
Sintered (α phase)	310	310	310	407	4.8	100-50
Reaction-sintered (20 percent free Si by volume)	380	415	275	345	4.4	100-50
SiC fiber-Si composite	275	275	275	340		70
CVD†	415	550	550	414		
			Silicon nitrides			
Hot-pressed (MgO additive)	690	620	330	317	3.0	30-15
Sintered (Y_2O_3 additive)	655	585	275	276	3.2	28-12
Reaction-bonded (2.45 g/cm^3)	210	345	380	165	2.8	6-3
β'-SiAlON (sintered)	485	485	275	297	3.2	22

*Room temperature. †Chemically vapor-deposited.

Reaction-formed SiC fiber–Si composites, developed by Hillig (*17*) at General Electric, are among the first engineered composite ceramic–ceramic structural materials. The process consists of starting with a graphite cloth, tow, felt, chopped fiber array, or any other possible precursor; forming a preform by any one of a variety of routes; and infiltrating the preform with liquid Si. The molten Si reacts with the filamentary graphite materials to form polycrystalline SiC fibers in a silicon metal matrix. The result is a fully dense oxidation-resistant body with about 30 to 50 percent Si (by volume) reinforced by 70 to 50 percent SiC fibers. (The high percentage of Si and the fibrous SiC morphology make this quite different from materials of the reaction-sintered SiC type.) Although this material shows promise of ease and versatility of fabrication, it is not yet clear what the production costs may be. The material permits the design of a composite component optimized for mechanical and thermal requirements. The presence of free silicon would be thought to limit use temperatures to ~ 1400°C. In spite of this, the material has been used in an experimental combustion liner application at temperatures above 1425°C. Although the material is still in a very early stage of development, the fact that it is an "engineered" composite material makes it a very exciting new development.

Chemically vapor-deposited SiC is a fully dense material with no additives. However, the strength of CVD SiC, which on occasion can be very high, is quite variable. This large scatter in strength is due to the occurrence of large columnar grains and residual deposition stresses. The strength values shown in Table 2 are from the lower end of the scatter band. While CVD SiC has been produced in complex shapes, the technique may be costly for production. The material may find eventual use as a coating for high-temperature oxidation and erosion resistance on SiC ceramics formed by more conventional routes.

Silicon Nitride Ceramics

Silicon nitride exists in two phases, α and β, both of which have hexagonal crystal structures. The α phase has a unit cell approximately twice as large as that of the β phase. The silicon nitride–based ceramics include hot-pressed, reaction-bonded, and sintered materials and the β'-SiAlON's (a solid solution of Al_2O_3 and/or other metal oxides in the β-Si_3N_4 lattice). Typical properties are given in Table 2.

Hot-pressed Si_3N_4 is produced by either conventional uniaxial or hot isostatic pressing. One starts with α-Si_3N_4 powder and adds a densification aid such as MgO, Y_2O_3, ZrO_2, or $SiBeN_2$. Under pressures of ~ 14 MPa and higher and temperatures of 1650° to 1750°C, some of the α-Si_3N_4 reacts with the additive and the thin layer of SiO_2 that coats each particle of Si_3N_4, producing a liquid silicate in which the remaining α-Si_3N_4 dissolves and from which it reprecipitates as elongated β-Si_3N_4 grains. On completion of the $\alpha \rightarrow \beta$ transformation, the elongated β grains are surrounded with a residual silicate or oxynitride grain boundary phase. The elongated nature of these grains, which are typically ~ 0.5 by ~ 4 micrometers, gives hot-pressed silicon nitride its high strength. Almost all hot-pressed silicon nitrides exhibit room-temperature flexural strengths of 690 MPa (100,000 psi) or higher and retain these strengths to at least 1000° to 1100°C, as well as exhibiting excellent thermal shock resistance and erosion and corrosion resistance.

Depending on the purity and phase composition of the starting Si_3N_4 powder, type and percentage of additive, milling and mixing procedures, and hot-pressing parameters (temperature, time, and pressure), one can obtain a wide variety of strength versus temperature, creep, or oxidation behaviors. Of all available high-temperature structural ceramics, hot-pressed Si_3N_4 with MgO as a densification aid (typified by the Norton Company's NC-132 material) most nearly approaches the reproducibility of mechanical properties expected of a true engineering material. However, the nature of the magnesium silicate grain boundary phase is such that these materials exhibit a rapid falloff in strength, creep, and oxidation resistance between 1200° and 1350°C.

Since it was apparent that the limitations of hot-pressed silicon nitride were due to the nature of the grain boundary phase and not intrinsic to the Si_3N_4 itself, Gazza and his co-workers in our laboratory (*18*) and Tsuge and his co-workers at Toshiba (*19*) focused attention on controlled modification of the grain boundary. This was an extremely difficult undertaking. There was no direct, but only inferential, evidence about the nature of the grain boundary chemistry and phase compositions. Most important, the grain boundary was thought to be only ~ 50 angstroms thick. [Subsequent direct observation of grain boundaries by Clarke and Thomas (*20*), using lattice imaging transmission electron microscopy (TEM), showed the boundaries to be only ~ 10 Å thick.] Since much of the grain boundary material is in the glassy state, even x-ray and lattice imaging TEM techniques often yield little information. To distinguish this emphasis on the grain boundary from the more usual interest in total microstructure, a "grain boundary engineering" approach was adopted in our laboratory. Tsuge and co-workers focused on "grain boundary crystallization." This approach has been particularly successful in developing hot-pressed Si_3N_4 with Y_2O_3 as a densification aid. Gazza (*21*) and others (*22*) demonstrated that Y_2O_3 additions provide higher strength at both room and elevated temperatures (to 1400°C) as well as better creep and oxidation resistance than obtained with Si_3N_4 containing MgO. However, this material has been plagued with an intermediate-temperature (~ 1000°C) oxidation problem. Although it appears that postfabrication heat treatment (*23*) and proper attention to composition and phase equilibria, as discussed by Lange (*22*), can alleviate this problem, more work needs to be done before hot-pressed Si_3N_4 with Y_2O_3 will be as reliable as the MgO-containing varieties. Strengths as high as 965 MPa (140,000 psi) at 1200°C have been obtained with a $Y_2O_3 + Al_2O_3$ additive, using the grain boundary crystallization approach (*24*). Through this focus on the grain boundary, the high-temperature strength of hot-pressed Si_3N_4 has been increased by an order of magnitude in less than 10 years. Nevertheless, hot-pressed Si_3N_4 is difficult to machine and components made from this material are relatively costly. While the high cost may be acceptable for some specialty applications, for more general use a lower-cost fabrication route must be developed.

Reaction-bonded Si_3N_4 has been developed largely to obtain a readily formable (little machining required) low-cost material (*25, 26*). In contrast to hot-pressed Si_3N_4, reaction-bonded Si_3N_4 maintains its strength to temperatures beyond 1400°C and exhibits significantly lower creep rates. As a consequence of the reaction-bonding process, this material is of necessity at least 10 percent porous, which makes it less oxidation-resistant than hot-pressed Si_3N_4 at intermediate temperatures, limits its strength to less than 415 MPa (60 kilograms per square inch) or more typically to about 245 MPa (35 ksi), and considerably reduces its erosion resistance.

Fabrication of reaction-bonded Si_3N_4 components begins with a silicon metal preform made by slip casting, dry pressing, flame spraying, injection molding, or various other techniques. The preform is then nitrided in an atmosphere of pure N_2 or $N_2 + H_2$. The nitridation of such a Si preform is a remarkable if still somewhat imperfectly understood phenomenon. For $3Si_{(s)} + 2N_{2(g)} \rightarrow Si_3N_{4(s)}$ there is a 23 percent expansion in the solid volume compared to Si, yet when this reaction is carried out on a preform there is essentially no change in dimensions (~ 0.1 percent). The reason for this appears to be that the first Si_3N_4 to form does so by a complex solid-liquid-vapor whisker growth into the void space of the Si preform, forming a skeleton-like structure, which in turn fixes the dimensions of the component. What appears to be a rather complex series of processing steps, and is a complex series of chemical reactions, yields a ceramic material with a unique property: it can be mass-produced to strict dimensional tolerances with little or no machining and at low cost.

Reaction-bonded Si_3N_4 has improved considerably over the past few years. Nevertheless, where high-strength, more oxidation-resistant material is required, it would be desirable to have a readily fabricable, fully dense Si_3N_4. This has been the impetus for the development of sintered Si_3N_4. However, reaction-bonded Si_3N_4 is still a viable candidate for many high-temperature applications and is currently under active development as an automotive turbocharger material in several laboratories.

Sintered silicon nitrides are a rather recent development. Although Si_3N_4 was sintered as early as 1973 (*27*), the balance between dissociation of the Si_3N_4 and densification during sintering was such that material of only about 90 percent theoretical density was obtainable. Indeed, the possibility of sintering fully dense Si_3N_4 was still an open question as late as 1976. Since then, several groups have succeeded in producing sintered Si_3N_4 of at least 95 percent density, and a few groups have obtained greater than 99 percent density. What occurred was that the progress being made by Prochazka and co-workers at General Electric in sintering SiC and other covalent compounds (*15*) was beginning to influence the thinking of the Si_3N_4 research community. Also, the concept of using a nitrogen overpressure and other techniques to suppress the density-limiting dissociation of Si_3N_4 at sintering temperatures was independently demonstrated by U.S. and Japanese investigators (*28, 29*). Finally, the experiences gained in using the grain boundary engineering or crystallization approach on hot-pressed silicon nitride was seen to be directly applicable to sintered material. As shown in Table 2, commercially available sintered Si_3N_4 has strengths falling between those of the hot-pressed and reaction-bonded materials.

Sintered Si_3N_4 has been formed by injection molding and, provided isotropic shrinkage can be obtained, components require little machining. Recently, Giachello and Popper (*30*), in a joint program of the Fiat Research Center and the British Ceramic Research Association, demonstrated that it is possible to post-sinter a reaction-bonded silicon nitride preform to ~ 98 percent theoretical density, with increased strength and oxidation resistance. Mangels and Tennenhouse (*31*) at Ford Motor Company have independently followed a similar line of research and have, in fact, fabricated components of sintered reaction-bonded silicon nitride. With this development one could start with a sintering preform that would yield only 6 to 8 percent linear shrinkage, as opposed to 18 to 20 percent linear shrinkage for sintered components. It is also possible that sintered Si_3N_4 bodies of more than 95 percent theoretical density may be used as preforms for hot isostatic pressing. Such a development would be a major breakthrough toward attaining high-reliability, affordable, high-performance components such as turbocharger rotors and diesel pistons.

SiAlON's represent an important new class of ceramic materials that are solid solutions of metal oxides in the β-Si_3N_4 crystal structure. These solid solutions produce a distorted β-Si_3N_4 lattice; hence, they are referred to as β'-SiAlON's. SiAlON's were originally developed with Al_2O_3, but MgO, BeO, Y_2O_3, and others have all been found to yield β' solid solutions, as well as a variety of other phases. Although SiAlON's were intended for application in heat engines, these materials are still in early development and have not been used in engine demonstration programs to date. It is likely that they will play some role in future engine programs. Since SiAlON phases are present in the grain boundaries of most hot-pressed or sintered Si_3N_4, they control the high-temperature behavior of these materials. Therefore understanding the phase relationships in these systems is of major importance. The bulk of phase equilibrium studies in these systems have been performed by Jack and his students at the University of Newcastle upon Tyne (*32*) and Tien of the University of Michigan with Gauckler of the Max Plank Institute and their co-workers (*33*).

This brief review of the silicon carbides and nitrides has focused on processing and the development of high-temperature strength. There are, of course, many other properties that must be optimized for these materials to be utilized in high-temperature engineering systems. Creep and oxidation resistance have been briefly referred to above and progress has been achieved in improving these properties. Currently much attention is being directed to the time dependence of strength in these high-temperature structural ceramics. Like all materials at elevated temperatures, they exhibit a decreasing ability to carry a given load as time increases. Data on the time-dependence of high-temperature strength of these materials is presently being accumulated for the purpose of life prediction calculations. As data on the mechanisms of time-dependent failure become available, it can be anticipated that ceramists will develop SiC's and Si_3N_4's with improved time-dependent strength. Certainly, the sections above suggest that, in general, if one can define a specific materials shortcoming, the materials research community will be able to design a material to overcome the problem. Let us now turn from design of ceramic materials back to design with ceramic materials and see how they are being applied.

Applications

As previously stated, there is considerable incentive to use high-temperature structural ceramics in a wide variety of systems. However, the incentive has been counterbalanced by the serious engineering problem of how to deal with the brittle nature of ceramics. Any experienced designer knows that you do not subject brittle materials to appreciable tensile stresses, even if the materials are reputedly strong. Quite appropriately, even the most innovative designer is not willing to accept a high-risk design. A decade ago the concept of using ceramics as highly stressed components in any system was decidedly a high-risk undertaking. Recognizing both the potential and the risk, and the need to integrate materials and design innovation in one systems-oriented program that would be structured to allow for several iterations of the systems design, the U.S. Department of Defense's Advanced Research Projects Agency (DARPA) initiated a Brittle Materials Design Program in 1971. A contract was awarded to Ford Motor Company, with Westinghouse Electric Company as a subcontractor and the Army Materials and Mechanics Research Center as monitor, to develop a design capability for brittle materials. This capability was to be demonstrated through the successful development and application of ceramic hardware in gas turbine hot-flow-path components to operate at 1373°C, uncooled, for 200 hours in a vehicular engine and for 100 peaking cycles in an electrical utility size turbine test rig.

The major underlying goal of the DARPA Brittle Materials Design Program was to demonstrate and encourage the use of ceramics as engineering materials. When the program was initiated, it was debatable whether ceramics could survive the rigors of the gas turbine environment in any meaningful capacity. During the course of this pioneering program, which has just ended, both reaction-bonded Si_3N_4 and reaction-sintered SiC stators, reaction-bonded Si_3N_4 nose cones and shrouds, as well as reaction-sintered SiC combustors have been demonstrated by Ford over duty cycle conditions (up to 1373°C, uncooled) in test rigs for more than 200 hours. A ceramic rotor fabricated from a hot-pressed Si_3N_4 hub bonded to a reaction-bonded Si_3N_4 blade ring was also demonstrated for 200 hours at 50,000 revolutions per minute and a gas temperature equivalent to operation at a turbine inlet temperature of 1200°C and a rotor rim temperature of 1000°C—beyond the capability of uncooled superalloys. Most significantly, an engine was run with a ceramic nose cone, stator, and rotor at temperatures between 1225° and 1415°C for 36 hours (*34*). This demonstrated an operating advantage for ceramics of about 300°C over what is possible with uncooled superalloys. As the DARPA program was demonstrating that ceramic components could survive in a gas turbine environment, other programs both in the United States and abroad were initiated to demonstrate ceramics in a variety of engine types. Although there are active programs on ceramic engine technology in West Germany, Japan, Britain, and Sweden, as well as the United States (*35*), I will review progress only on several selected U.S. programs.

In the area of ceramics for gas turbine application, there have been three significant milestones within the past year. Under a joint DARPA-Navy program (*36*), Garrett Corporation has demonstrated the integration of more than 100 separate ceramic components into a modified, ceramic-configured (two stages of ceramic stator vanes and two stages of ceramic-bladed rotors) turboshaft engine, operation of the engine for several hours, and attainment of a more than 30 percent increase in horsepower. Although this engine still has many development problems, it has provided the first demonstration that, even with the design compromises required for their incorporation into an engine, ceramics can deliver the increased performance that their advocates have promised.

The second major event was the road and test track demonstration of a partially ceramic-configured Detroit Diesel Allison 404 gas turbine engine in a truck, under Department of Energy sponsorship. In this demonstration an engine with reaction-sintered SiC stator vanes and aluminosilicate ceramic regenerators was installed in the truck after ~ 1800 hours of laboratory testing. The truck was then driven from Indianapolis to Detroit and subjected to about 25 hours of shock and vibration testing as well as about 100 hours of road tests (*35*). In the third achievement, the Solar Division of International Harvester, under contract from the U.S. Army Mobility Engineering Research and Development Command, conducted a successful, full-performance, 200-hour demonstration of an all-ceramic nozzle section (hot-pressed Si_3N_4 vanes with reaction-sintered SiC shrouds) in a 10-kilowatt turbogenerator. This is the first known instance of a ceramic-configured engine producing electrical power (*35*).

Progress has been made in the diesel engine area as well. Perhaps the most exciting advanced concept in diesel technology is the so-called adiabatic turbocompounded diesel. This engine would eliminate the water-cooling system, and the heat (energy) that would have been rejected to the cooling system is diverted to the exhaust, recovered via a turbine, and geared back into the engine output shaft. Fully developed, such an engine can reduce fuel consumption by as much as 25 percent, with a significant increase in specific power and a decrease in the maintenance problems associated with the water-cooling system. To attain such gains, ceramic pistons or piston caps, cylinder liners, exhaust gas manifolding, and valve train components are required. The U.S. Army Tank and Automotive Research and Development Command, jointly with Cummins Engine Company, is currently developing such an engine (*9*). The most highly stressed component in such an engine is the piston cap. An important milestone, therefore, is the demonstration of a ceramic piston cap in an engine. Such a demonstration was successfully completed in a 250-hour performance test in a single-cylinder engine during the past year, utilizing a hot-pressed-to-shape Si_3N_4 piston cap fabricated in our laboratory (*35*).

Ceramic components are under active development for other applications in piston engines, as turbochargers, diesel combustion prechambers, valve lifters, wrist pins, and push rod tips. The incentives for the use of ceramics in these applications are not high-temperature capability but rather lower cost, improved wear capability, increased stiffness, and reduced inertia for improved response and performance. Carborundum, AiResearch Casting Company, and Ford Motor Company among others, are actively pursuing development of Si_3N_4 and SiC turbochargers. Carborundum has sintered SiC valve lifters under test in racing cars, and thus far they are providing outstanding performance and durability.

In another area, Hague International is commercially marketing a high-temperature heat exchanger for use on slot forging furnaces (maximum temperature ~ 1250°C). What is impressive in this application is that the heat exchange tubes are finned SiC elements about 50 inches long and that they operate successfully in the shock and vibration environment of a forge furnace (*10*). The corrosion and erosion resistance of SiC and Si_3N_4 ceramics, in particular, make them likely to be used as heat exchangers and combustors in the indirect firing of turbines with coal or heavy residual oil (*3*). A future application for Si_3N_4- or SiC-based heat exchangers would be in the solar energy cavity receiver for solar "power towers" operating hot-air turbines.

Aside from the high-temperature engineering applications for the SiC and Si_3N_4 materials, there are also potential applications in the metals processing, wear-resistant materials, and bearing areas. Hot-pressed Si_3N_4 in particular has shown significant potential for use in high-performance ball and roller bearings (*37*). Work on hot-pressed Si_3N_4 in the People's Republic of China (*38*) and on SiAlON in Britain (*39*) has shown that these materials give good performance as tool bits in machining selected materials.

The foregoing discussion clearly demonstrates that high-performance ceramics technology has come a long way in the past 10 years. We have gone from discussing whether ceramics can survive in an engine to actually having ceramics in engines on the highway, on the racetrack, and producing electrical power. Yet there is still a long way to go. The applications discussed above are for the most part one-of-a-kind demonstrations. For most applications ceramic components are not yet ready for introduction into commercial engines or other high-performance systems. What will it require for this technology to become commercialized?

Outlook

For high-performance ceramic materials to be successfully commercialized in mass-production applications such as automotive engines, several key problems must be solved. Primary among these is ceramic attachment. The section above described enough successes of ceramic application in engines to show that considerable progress has been made in the attachment area. What I did not mention above is that virtually every case of a ceramic component failure in an engine or a test rig, that I am aware of, has been the result of an attachment problem. Thus, advanced design and materials concepts for minimizing contact stresses in ceramic-to-ceramic and ceramic-to-metal attachments maintaining compliance and integrity at high temperatures and for long times constitute a crucial R & D area. The second most critical question is whether the relatively few components demonstrated to date can be scaled up to quantity production reliably, reproducibly, and affordably. This is a problem that the ceramics industry deals with each time a new technology using ceramics is introduced. Recent examples of reliable high-volume manufacturing of high-performance ceramics such as alumina substrates, envelopes for sodium vapor lamps, SiC igniters for the gas appliance industry, and ZrO_2 for automotive emissions sensors give reason for optimism that the levels of product uniformity required for industrial application can be attained. The third key area is that of nondestructive testing (NDT). Significant increases in NDT and other quality assurance methods will be required to ensure that only reliable components are put into use. At present, this is beyond the state of the art for the materials and applications discussed above. These current problems notwithstanding, it is clear that these materials are serious candidates for application in several areas of engine technology.

Applications such as valve lifters, exhaust port liners, and diesel precombustion chambers are fairly near-term ones that will not require major changes in production lines or engine technology. The same is true of the turbocharger applications, although here some material improvement and attachment work is required, so it is probably not quite so close. Ceramic heat exchangers for industrial waste heat recovery are already commercially available for some applications and will gradually spread into a variety of areas. The use of ceramics in large diesel engines through the adiabatic turbocompound engine is a good likelihood for commercialization by the 1990's. Ceramic components in truck and small industrial gas turbine engines can be phased in gradually and are also likely to be commercialized. The major question is whether the gas turbine, with ceramic components, will be adapted as an alternative engine for the automotive industry. The Department of Energy's Advanced Automotive Gas Turbine Program (*11*) is currently funding the development and evaluation of two such engines, one by AiResearch Company with Ford Motor Company as a subcontractor, and one by Detroit Diesel Allison in conjunction with Pontiac. These engines integrated into vehicles are due for demonstration around 1982 to 1984. Assuming technical success of these programs, there is still the issue of the investment costs for this new technology versus the costs of advanced piston engine technology. Conversion of the nation's automotive engine production lines is a multibillion-dollar undertaking. Whether ceramic gas turbine technology is fully utilized will depend on complex cost-benefit trade-offs between it and alternative technologies. What is important in our increasingly energy- and resource-scarce world is that such options are at least available. In any event, over the next decade we will begin to see high-temperature ceramics utilized in applications that would have been unimagined a decade ago.

References and Notes

1. *Should We Have a New Engine?—An Automotive Power Systems Evaluation* (Rep. JPL-SP 43-17, Jet Propulsion Laboratory, Pasadena, Calif., 1975), vols. 1 and 2.
2. J. J. Burke, E. M. Lenoe, R. N. Katz, Eds., *Ceramics for High Performance Applications—II* (Brook Hill, Chestnut Hill, Mass., 1978).
3. *Proceedings of the Workshop on Ceramics for Advanced Heat Engines* (Rep. CONF 770110, ERDA, Washington, D.C., 1977).
4. *Proceedings of the Conference on Basic Research Directions for Advanced Automotive Technology* (Department of Transportation, Washington, D.C., April 1979).
5. R. W. Davidge, in *Nitrogen Ceramics*, F. Riley, Ed. (Noordhoff, Leyden, 1977), pp. 653-657.
6. J. L. Klann, *Advanced Gas Turbine Performance Analysis* (8th Summary Rep., Automotive Power Systems Contractor's Coordination Meeting, ERDA-64, May 1975), p. 173.
7. H. E. Helms and F. A. Rockwood, *Heavy Duty Gas Turbine Engine Program Progress Report* (Rep. DDAEDR 9346, contract NAS3-20064, NASA-Lewis Research Center, February 1978).
8. F. B. Wallace *et al.*, in (*2*), pp. 593-624.
9. R. Kamo, in (*2*), pp. 907-922.
10. R. A. Penty and J. W. Bjerklie, in *New Horizons Materials and Processes for the Eighties, Proceedings of the 11th National SAMPE Technical Conference* (SAMPE, Azusa, Calif., 1979); R. A. Penty, personal communications.
11. G. Thur, paper presented at the 17th Department of Energy Highway Vehicle Systems Contractors' Coordination Meeting, Dearborn, Mich., October 1979.
12. G. Q. Weaver and B. A. Olson, in *Silicon Carbide—1973*, R. C. Marshal, J. W. Faust, Jr., C. E. Ryan, Eds. (Univ. of South Carolina Press, Columbia, 1974), pp. 367-374.
13. S. Prochazka, in *ibid.*, pp. 391-402.
14. R. J. Bratton and D. G. Miller, in (*2*), p. 719.
15. S. Prochazka, in *Ceramics for High Performance Applications*, J. J. Burke, A. E. Gorum, R. N. Katz, Eds. (Brook Hill, Chestnut Hill, Mass., 1974), pp. 239-252.
16. J. A. Coppola and C. H. McMurty, in *National Symposium on Ceramics in the Service of Man* (Carnegie Institution of Washington, Washington, D.C., 1976).
17. W. B. Hillig, in (*2*), pp. 979-1000.
18. R. N. Katz and G. E. Gazza, in (*5*), pp. 417-431.
19. A. Tsuge, K. Nishida, M. Komatsu, *J. Am. Ceram. Soc.* **58**, 323 (1975).
20. D. R. Clarke and G. Thomas, *ibid.* **60**, 491 (1977).
21. G. E. Gazza, *Am. Ceram. Soc. Bull.* **54**, 778 (1975).
22. R. J. Bratton, C. A. Anderson, F. F. Lange, in (*2*), pp. 805-825.
23. G. E. Gazza, H. Knoch, G. D. Quinn, *Am. Ceram. Soc. Bull.* **57**, 1059 (1978).
24. A. Tsuge and K. Nishida, *ibid.*, p. 424.
25. D. J. Godfrey, *J. Br. Interplanet. Soc.* **22**, 353 (1969).
26. D. R. Messier and P. Wong, in (*15*), pp. 181-194; J. Mangels, in *ibid.*, pp. 195-206.
27. G. R. Terwiliger and F. F. Lange *J. Mater. Sci.* **10**, 1169 (1975).
28. H. F. Priest, G. L. Priest, G. E. Gazza, *J. Am. Ceram. Soc.* **60**, 81 (1977).
29. M. Mitomo, M. Tsutsumi, E. Bannai, T. Tanaka, *Am. Ceram. Soc. Bull.* **55**, 313 (1976).
30. A. Giachello and P. Popper, paper presented at the 4th International Meeting on Modern Ceramic Technologies, St. Vincent, Italy, 28 May to 1 June 1979.
31. J. A. Mangels and G. J. Tennenhouse, *Am. Ceram. Soc. Bull.* **58**, 884 (1979).
32. K. H. Jack, in (*5*), pp. 103-128.
33. L. J. Gauckler, S. Boskovic, G. Petzow, T. Y. Tien, in (*2*), pp. 559-572.
34. A. F. McLean and J. R. Secord, *U.S. Army Mater. Mech. Res. Cent. Rep. TR 79-12* (1979).
35. Presentations on these programs were made at the Conference on Ceramics for High Performance Applications III—Reliability, held at Orcas Island, Wash., July 1979.
36. W. L. Wallace, J. E. Harper, F. B. Wallace, "Ceramic Gas Turbine Engine Demonstration Program," Interim Rep. 13 on contract N00024-76-C-5352, May 1979.
37. C. F. Bersch, in (*2*), pp. 397-406.
38. H.-C. Miao, Y.-H. Liu, T.-C. Chiaug, paper presented at the 4th International Meeting on Modern Ceramic Technologies, St. Vincent, Italy, 28 May to 1 June 1979.
39. R. J. Lumby, B. North, A. J. Taylor, in (*2*), pp. 893-906.

Aircraft Gas Turbine Materials and Processes

B. H. Kear and E. R. Thompson

For more than three decades the development of the gas turbine engine has been paced by the availability of materials and the ability to process them into useful shapes. The most challenging materials problems have been encountered in aircraft gas turbines. This is because of the need to maintain high operating efficiencies without incurring unacceptable weight penalties. It is to the credit of materials technologists that these challenges have continued to be met, as engine designs have progressed to ever-increasing levels of engineering sophistication and performance. As an indication of the remarkable progress that has been made over the years, it may be noted that, since the 1950's, thrust-to-weight ratios have tripled, fuel efficiencies have more than doubled, and the time between overhauls has increased from 100 to more than 10,000 hours.

The most significant developments in alloy design occurred in the early 1950's. Entirely new classes of heat-resistant nickel- and cobalt-base alloys were developed which became known later as the superalloys. At the same time, a new class of titanium alloys of high specific strength became available in usable structural forms. The superalloys proved to be of great utility in the hottest parts of the engine, such as burner and turbine sections, whereas the titanium alloys were ideal for the cooler compressor section of the engine. The effect of these developments was to cause a sharp increase in the use of superalloys and titanium alloys in engines, at the expense of conventional nickel- and iron-base alloys (Fig. 1).

Starting in the mid-1960's, the emphasis gradually shifted from alloy development toward process development. In the productive period that followed, several important advances were made in the materials processing area. Perhaps the most striking innovation was the introduction of directional solidification processing of turbine blades and vanes.

B. H. Kear is a senior consulting scientist and E. R. Thompson is a group leader at United Technologies Research Center, East Hartford, Connecticut 06108.

0036-8075/80/0523-0847$00.50/0