(NASA-TM-111137) COMPOSITES RESEARCH AT NASA LEWIS RESEARCH CENTER (NASA. Lewis Research Center) 24 p

N96-14667

Unclas

G3/24 0077633

Composites Engineering, Vol. 4, No. 8, pp. 787-810, 1994
Elsevier Science Ltd
Printed in Great Britain

NDB
NASA
IN-24-TM
6100
P-24

0961-9526(94) E00031-X

COMPOSITES RESEARCH AT NASA LEWIS RESEARCH CENTER

Stanley R. Levine, Stephen Duffy,† Alex Vary, Michael V. Nathal, Robert V. Miner, Steven M. Arnold, Michael G. Castelli, Dale A. Hopkins and Michael A. Meador

National Aeronautics and Space Administration, Lewis Research Center, Cleveland, OH 44135, U.S.A.

(*Received* 25 *March* 1994; *final version accepted* 29 *April* 1994)

Abstract—Composites research at NASA Lewis is focused on their applications in aircraft propulsion, space propulsion and space power, with the first being predominant. Research on polymer-, metal- and ceramic-matrix composites is being carried out from an integrated materials and structures viewpoint. This paper outlines some of the topics being pursued from the standpoint of key technical issues, current status and future directions.

1. INTRODUCTION

Advanced composites are a key to the development of the next generation of civil transport aircraft engines. The driving forces for the development of advanced engines are mission-enabling capabilities and reduced life-cycle costs. An example of a mission-enabling capability is an advanced environmentally friendly engine for a 300 passenger supersonic civil transport intended for entry into service in about 2005. Requirements for this engine include NO_x emissions less than 5 gm kg^{-1} of fuel burned, noise emissions in compliance with FAR 36 Stage III, and low engine weight and acceptable engine performance to make such an aircraft economically attractive to the customer (Stephens *et al.*, 1993). Composites, and other advanced materials, will play a key role in meeting these goals as well as providing us with more economical and efficient subsonic air transportation. An example of potential composites applications in a highly advanced high-bypass ratio turbofan for a subsonic transport is shown in Fig. 1 (Stephens, 1990).

High-temperature composites research at NASA Lewis Research Center is primarily focused on aircraft engines. The effort includes both materials and structures research addressing the materials classes illustrated in Fig. 1, i.e. polymer-matrix composites, metal- and intermetallic-matrix composites, and ceramic-matrix composites. Of necessity, our concerns include constituent development and property characterization; composite fabrication and process modeling; nondestructive evaluation; constituent and composite property models and design codes; and prediction and measurement of performance and life under actual or simulated engine conditions. The purpose of the paper is to provide the reader with an overview of many, but not all, of our composites research efforts. In the brief sections which follow, we will address each area in terms of its key technical issues, current status and future directions.

2. POLYMER MATRIX COMPOSITES‡

The use of PMCs in aircraft engines can result in significant weight savings and lead to improved fuel economy, increased payload or increased flight distances. However, the poor thermal and thermal oxidative stability of these materials limits their use to the cooler sections of the engine. Considerable advances have been made over the years to improve the stability of PMCs so that current materials can tolerate extended use at temperatures up to 650°F. While PMCs are the most mature of all composite materials,

† Cleveland State University.
‡ Contributed by Michael A. Meador.

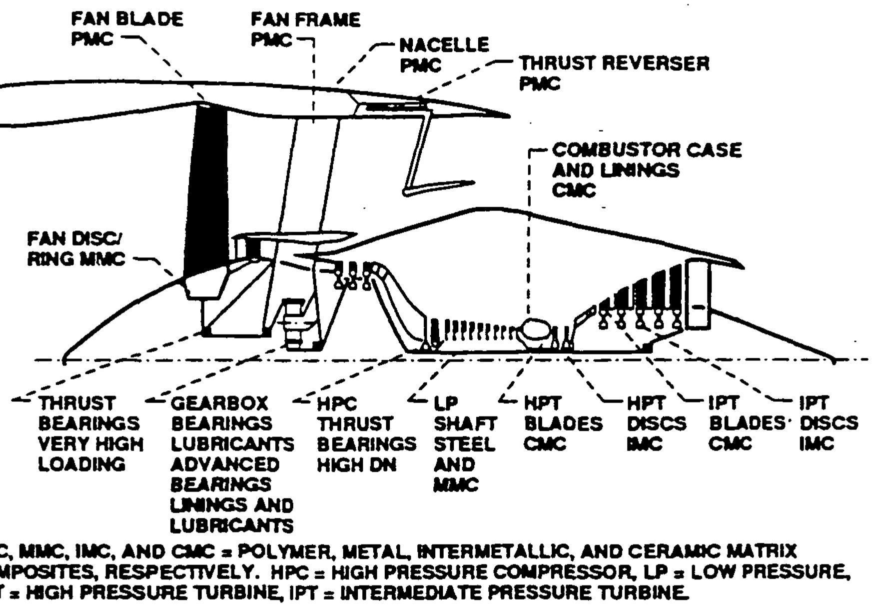


Fig. 1. Advanced materials applications for an ultra-high-bypass engine (Stephens, 1990).

a number of challenges need to be conquered before they can be fully utilized in both commercial and military aircraft engines. Among these are long-term durability, processability, affordability and repairability.

Over the past two decades, research at the NASA Lewis Research Center has primarily dealt with improving the stability and processability of high-temperature PMCs (Meador *et al.*, 1990). Processability and stability are often mutually exclusive properties for PMCs. Improved thermal oxidative stability in polymers is commonly achieved through the use of stable aromatic groups, e.g. benzene rings. These aromatic groups are rigid, highly planar structures and often render the polymers from which they are made difficult to melt and intractable.

A balance between processability and stability can be achieved by the use of the PMR approach developed at NASA Lewis in the early 1970s (Fig. 2). Molecular weight, and, hence, melt flow and processability, is controlled through the use of a latent reactive endcap (or chain terminating group). At high temperature, this endcap undergoes a cross-linking reaction to provide a material with good stability, high glass transition

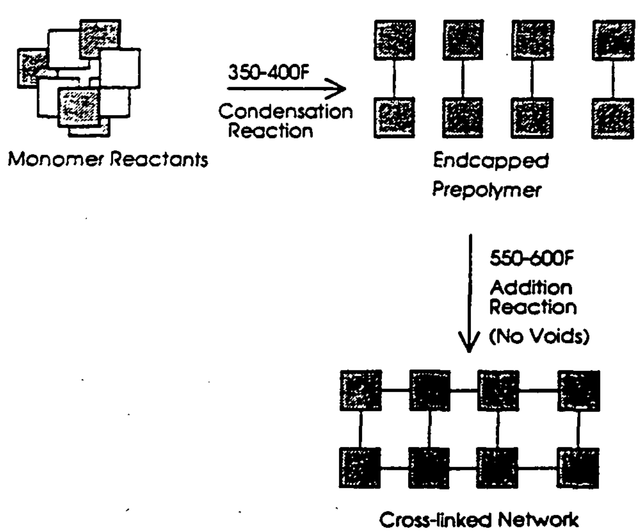


Fig. 2. Reaction scheme for addition polyimides.

temperature, and good mechanical properties (both at room temperature and elevated temperatures). The first polymer developed using this approach was PMR-15 (Serafini *et al.*, 1972), a high-temperature polyimide capable of extended use at temperatures below 500°F.

A variety of high-temperature polyimides have been prepared via this approach (Meador *et al.*, 1991) in an attempt to develop new materials with better stability than PMR-15. Second-generation PMR polyimides, PMR-II, were developed by substituting more thermal-oxidatively stable monomers into the polyimide backbone; this resulted in a 50°F increase in the upper-use temperature (Serafini *et al.*, 1976). Further modifications produced high-molecular-weight versions of PMR-II which showed potential for use at temperatures up to 600°F (Vannucci, 1987).

Recently, efforts have been directed at improving the thermal oxidative stability of PMR-II polyimides via endcap substitution. This has led to the development of styrene-endcapped polyimides, V-CAPs (Vannucci *et al.*, 1990) and paracyclophane-capped polyimides, CyCAPs (Waters *et al.*, 1991). Both systems have better processability than high-molecular-weight versions of PMR-II and can be used at temperatures as high as 650°F.

While these modifications of PMR-15 have resulted in new polyimides with better thermal-oxidative stability, this has been achieved with some sacrifice in processability. This is primarily due to the fact that these formulations have molecular weights 3–5 times that of PMR-15. This results in polyimides with melt viscosities nearly three orders of magnitude higher than PMR-15! Recent efforts have focused on reducing melt viscosities via (1) monomer substitution and (2) reduced cross-link density.

The melt viscosities of polyimides and other polymers can be reduced and their melting points lowered by altering their molecular structure to inhibit crystal packing and other intermolecular interactions in the solid. Since the chemical structures of most polyimides are fairly linear, crystal packing in these systems can be disrupted by using monomers with twists, kinks or other flexible linkages. Considerable reductions in the melt viscosities of PMR-II and V-CAP polyimides can be achieved through the use of a 2,2′-trifluorobenzidine, a diamine with a twisted or noncoplanar geometry (Chuang

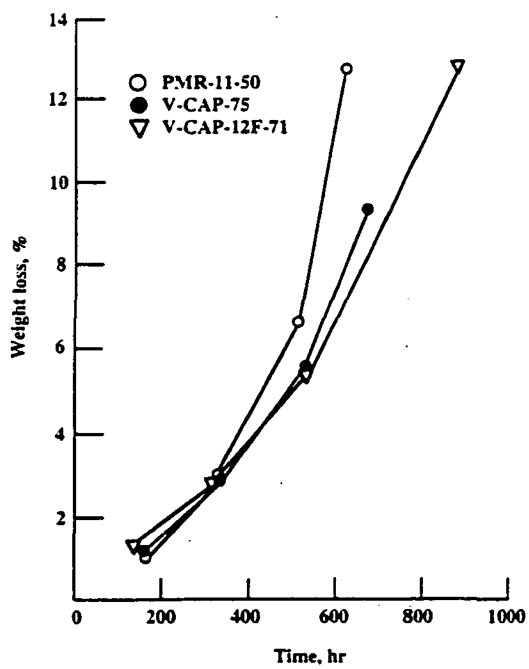


Fig. 3. A comparison of the thermo-oxidative weight losses of G-40-600 reinforced composites prepared with PMR-II-50, V-CAP-75, and a 2,2-bistrifluoromethylbenzidine substituted V-CAP (V-CAP-12F-71) after aging in 1 atm air at 371°C (700°F).

et al., 1992). Polyimides prepared with this diamine have better thermal-oxidative stability than both PMR-II and V-CAP resins (Fig. 3). Another twisted biphenyldiamine has been used to prepare thermoplastic polyimides which have the potential for use in high-temperature synthetic fibers (Chuang *et al.*, 1994). Reduced melt viscosities have also been achieved through the use of a series of flexible multi-ring diamines (Delvigs *et al.*, 1994). However, due to the presence of oxidizable methylene groups, polyimides made with these diamines have stabilities comparable to that of PMR-15.

Reduced cross-link density in addition-cured polyimides can also improve processability; however, this may result in decreased glass transition temperatures and poorer thermal-oxidative stability (Vannucci *et al.*, 1992).

Recent concern over the use of mutagenic diamines, such as methylenedianiline (U.S. Department of Health and Human Services, 1986), has spurred a considerable amount of activity aimed at protecting workers and the environment from potential health risks posed by the use of toxic or carcinogenic diamines. This has led to the search for diamines which do not pose a health risk to materials suppliers and fabricators. A variety of diamines have been examined as replacements for MDA in PMR-15. However, many of these do not provide polyimides with acceptable mechanical properties and thermal-oxidative stability. Polyimides prepared with a mixture of some of these diamines show some promise as MDA replacements (Vannucci and Chriszt, 1993).

The overall performance and durability of PMCs is strongly influenced by the strength of the resin–fiber interface (Bowles, 1990). A variety of graphite fibers are commercially available today. The method of preparation and the surface treatment of each of these fibers is different and information on these processes is usually proprietary. A recent study on a composites prepared with a series of commercially available graphite fibers shows that the nature and strength of the resin–fiber interface is strongly influenced by dipolar interactions occurring between the resin and fiber surface (Serrano *et al.*, 1994) (Fig. 4). These dipolar interactions occur between polar functional groups (hydroxyl, carbonyl and carboxylic acid) present on the fiber surface and the polyimide chain. A strong correlation was found between the polar energy of the fiber surface (measured by both fiber wetting and XPS) and the interlaminar shear strength of PMR-15 composites reinforced with that fiber. A similar correlation was found between composite weight loss and fiber surface polar energy. More work is needed to better characterize these dipolar interactions in order to tailor the fiber surface to improve the strength of the resin–fiber interface.

Oxidation-resistant coatings can also improve the thermal oxidative stability of PMCs. A variety of ceramic coatings have been applied to PMR-15 and PMR-II composites via plasma-assisted chemical-vapor deposition. Silica coatings up to 3500 Å thickness applied to a PMR-15 composite substrate reduced weight losses after 300 h aging in air at 390°C from 20% to nearly 5% (Miller and Gulino, 1994). A five-fold reduction in the weight loss of a PMR-II-50 composite after 300 h at 371°C was achieved with the use of a silicon nitride coating (Harding and Sutter, 1993). This coating survived 1000 thermal cycles from 25 to 371°C without any signs of cracking (Fig. 5). While all of these coatings adhere well to resin-rich surfaces, they do not adhere well to the machined surfaces of composites. Since engine components have bolt holes and machined surfaces, this problem must be solved before oxidation-resistant coatings can be used on PMCs in these applications.

Research in the Polymers Branch at the NASA Lewis Research Center has attempted to overcome some of the technical challenges that prevent the effective utilization of PMCs in aircraft engines. The long-term durability and stability of PMCs can be improved through the use of more stable resin systems, through better understanding and control of the resin–fiber interface, and through the use of oxidation-resistant coatings. Processability in high-temperature polymers can be enhanced by controlling and modifying the polymer's molecular structure. However, further work is needed to develop materials and processes which decrease the manufacturing costs and improve the reliability of high-temperature PMC components. These areas are currently under investigation.

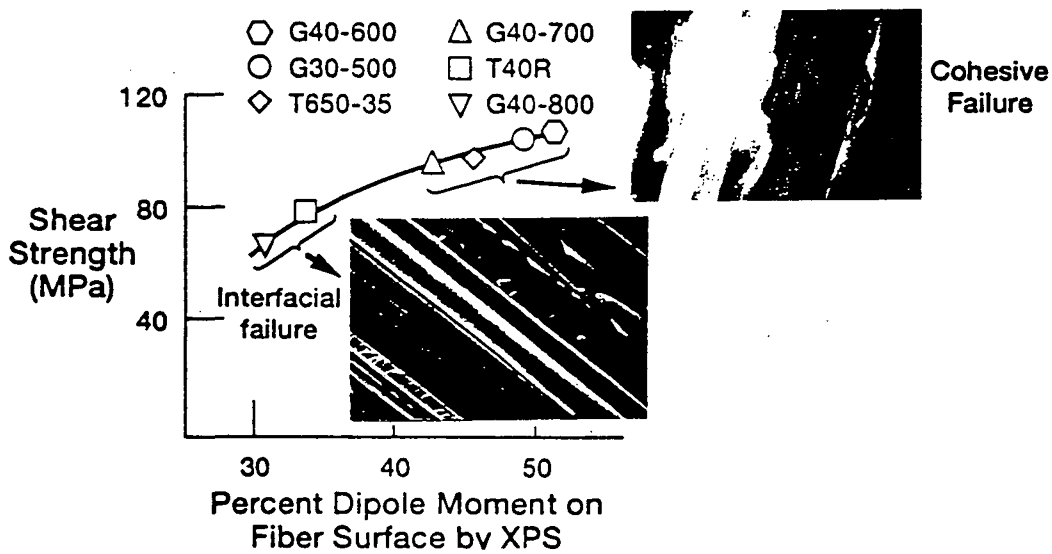


Fig. 4. The effects of fiber surface polarity (per cent dipole moment) on the interlaminar shear strength of PRM-15 composites reinforced with a variety of graphite fibers.

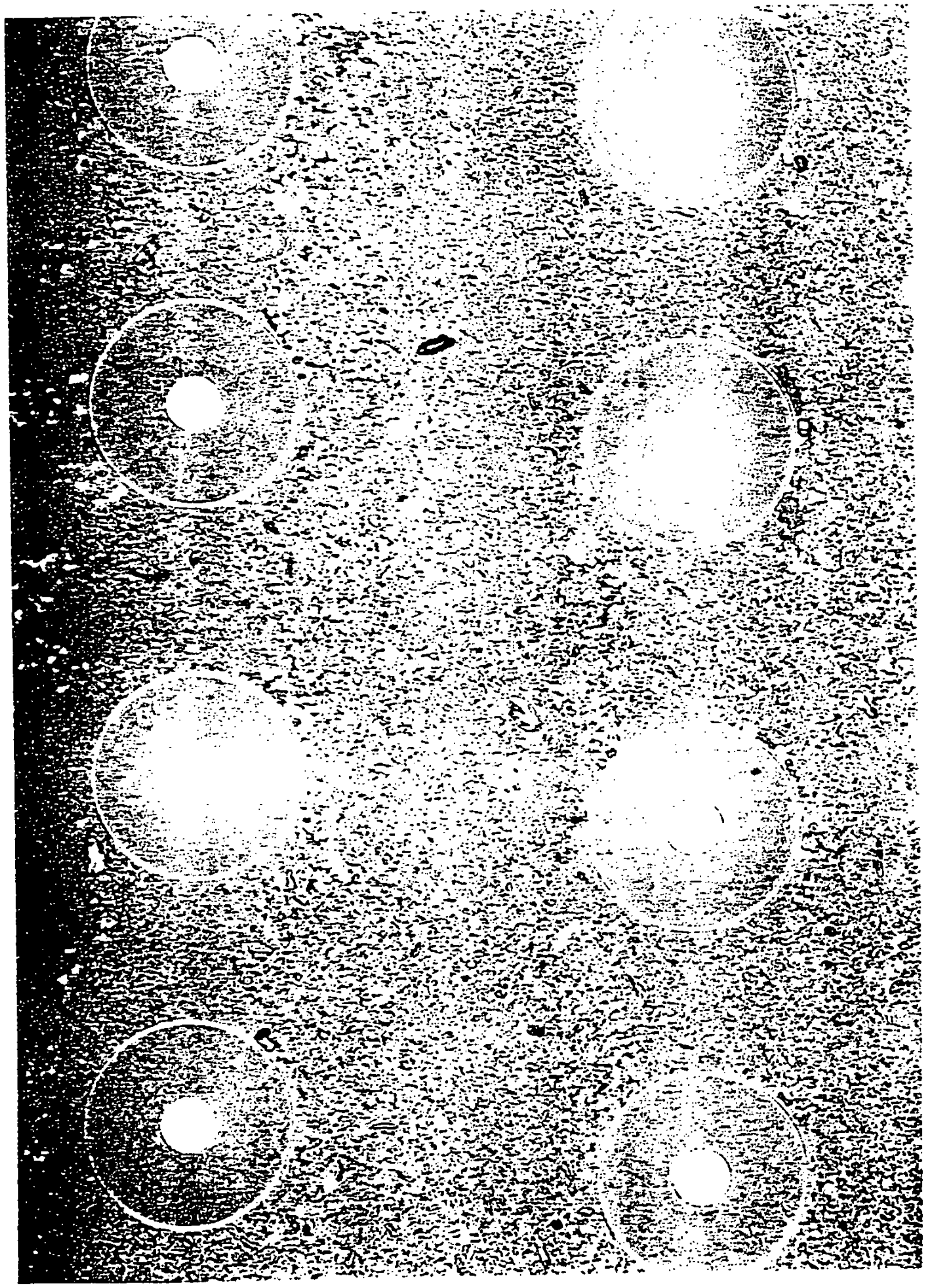


Fig. 9. SCS-6 SiC fiber/reinforced SiC matrix composite formed by reaction forming. The major dispersed matrix phase is $NbSi_2$ (Singh *et al.*, 1994).

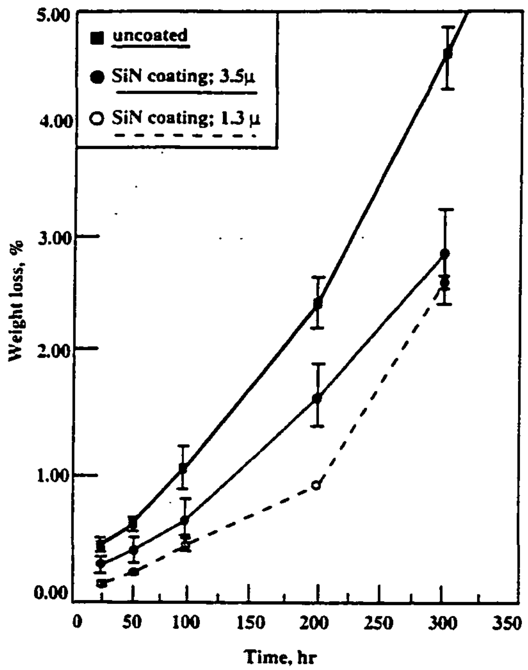


Fig. 5. A comparison of the thermo-oxidative weight losses of uncoated and silicon nitride-coated T-650-42-reinforced PRM-II-50 composites after aging in 1 atm air at 371°C (700°F).

3. METAL-MATRIX COMPOSITES†

3.1. *Materials research*

Metal-matrix composites (MMCs), which are defined here to include intermetallic matrices, have received considerable attention as candidates for advanced aerospace applications. These include advanced military and commercial aircraft, the supersonic High-Speed Civil Transport (HSCT), the National Aerospace Plane, and several applications in rocket engines such as those used on the Space Shuttle. These materials can offer higher strength and stiffness at lower weight than current monolithic Ti and Ni alloys. Research at NASA Lewis has focused on Ti-based MMCs for applications in the compressor section of commercial subsonic aircraft, and Fe-and Ni-based MMCs for applications in the turbine section of subsonic aircraft and the exhaust nozzle of the HSCT.

The first MMCs in aircraft gas turbine engines will likely be a Ti-based MMC used in a low-risk static part, but the highest payoffs will be attained with rotating parts such as reinforcing rings in compressor disks. One key issue for these composites is the need for manufacturing technology to produce reliable components at a reasonable cost. Ti-MMCs incorporating SCS-6 fibers produced by the foil/fiber/foil, arc-spray, plasma-spray and powder-cloth processes by various organizations have roughly comparable properties (MacKay *et al.*, 1991; Pickens *et al.*, 1993). Presumably the same may be true for tape casting; however, results have not been published. Because fiber strength dominates the 0° composite strength, distinctions between processes must be made using the same—or an equivalent—strength fiber lot. One of few such studies made has found equivalent 0° strength between composites made by powder cloth and plasma spray (MacKay *et al.*, 1994), two very dissimilar processes. This observation was explained by the fact that equivalent strengths were measured in fibers extracted from the composites, even though plasma spraying had produced some exfoliation of the carbon coating.

Other advantages and disadvantages of these processes in terms of cost, off-axis properties and long time durability are still being assessed. The foil/fiber/foil process is the most mature. It and the arc-spray process, which requires matrix alloy wire, promise lower oxygen levels than the processes using powder, but are limited to formable alloys

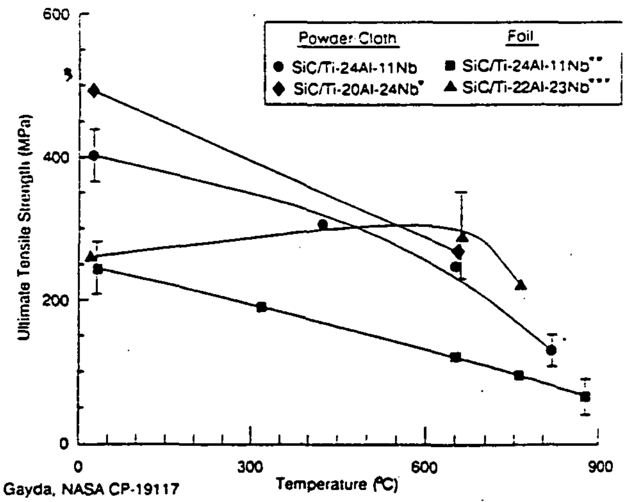
† Contributed by Robert V. Miner and Michael V. Nathal.

and will probably result in higher cost. The two thermal-spray processes, arc and plasma spray, share an advantage in uniform fiber placement, but possible long-time fiber degradation due to some exfoliation of the carbon coating must be explored. Of the powder-based processes, both tape casting and the powder-cloth method utilize polymer binders that must be removed prior to final consolidation, although impurity levels in laboratory coupons have been equal to or lower than those measured in MMCs made by competing processes (MacKay *et al.*, 1994). Tape casting is probably to be favored over the powder-cloth method as a more continuous process. However, yields may be lower and oxygen levels may be higher since a finer powder fraction is required.

Key property issues which will limit the range of application for Ti-MMCs are their environmental resistance and transverse properties. Their low environmental resistance is accentuated in thermomechanical fatigue (TMF) loading. The Ti-MMCs studied to date all exhibit very low lives in TMF cycles having tensile loading at low temperatures (Gabb *et al.*, 1993). TMF behavior in air will likely restrict use to temperatures below about 500°C. Transverse properties of Ti-MMCs are usually found to be lower than those of the monolithic matrix. Transverse tensile strength and ductility (Brindley and Draper, 1993), fatigue resistance in both air and vacuum (Gayda and Gabb, 1992; Lerch, 1990) and TMF resistance (Castelli, 1993) all show this trend. In these studies, it has been shown that failure initiates by debonding of the fiber from the matrix, which can occur in one or more of the C-rich layers of the SCS-6 coating and/or the reaction zone. The limited transverse properties may be overcome by cross-plied fiber architectures, which has been successful in some but not all cases (Lerch, 1990; Larsen *et al.*, 1992). As shown in Fig. 6, the alternative strategy of improving matrix composition also can be very effective in improving transverse composite strength. Both matrix strength and ductility are considered important in determining composite strength, although composite ductility has remained low (Brindley and Draper, 1993).

Despite the property limitations which have been found for Ti-MMCs, applications such as reinforcing rings in the bore of compressor disks show considerable promise because temperatures are limited, direct contact between the MMC and oxygen is excluded, and transverse loads are low. Such components have been successfully engine tested (Kandebo, 1992).

In contrast, the technology for (Fe,Ni)-based MMCs is much less mature, as it is still in the stage of laboratory-scale coupons. These composites have potential to operate in the 1000–1100°C range as turbine and nozzle components. Oxide fibers, particularly Al_2O_3, remain as the best current choice for reinforcement, due to their more favorable



*Gabb and Gayda, NASA CP-19117
**Gambone, WRDC-TR-89-4145
***Smith et al., WL-TR-92-4035

Fig. 6. Transverse properties of several SCS-6 reinforced titanium aluminide MMCs. By varying composite processing methods and especially matrix composition, the transverse tensile strength has been doubled in the last few years. After Brindley and Draper (1993); reproduced by permission of the U.S. Government.

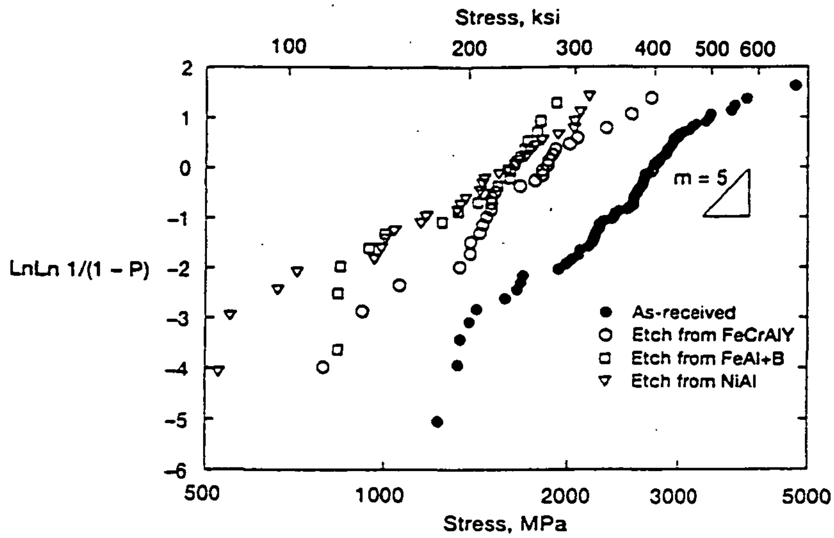


Fig. 7. Weibull probability plots showing strength degradation of sapphire fiber after composite consolidation (Draper and Locci, 1994).

chemical and thermal expansion compatibility with the matrix compared to SiC fibers. Although intermetallics such as NiAl offer the greatest potential due to their light weight and higher-temperature capability (Bowman, 1992), our current focus is on utilization of the more ductile superalloys as nearer-term matrices. Even in the more ductile superalloy matrix composites, however, significant technical challenges need to be resolved. Of prime importance is the need to prevent fiber-strength degradation in Al_2O_3 fibers. Sapphire, the Al_2O_3 fiber with the highest strength potential, has been shown to suffer from strength reductions of the order of 50% when exposed to a variety of matrices at typical composite consolidation cycles (cf. Fig. 7; Draper and Locci, 1994). This strength degradation appears to be caused by the introduction of surface flaws, although degradation has been observed even when the extent of reaction between fiber and matrix is very slight or completely absent. Fiber coatings have been proposed as a solution to this problem. Other oxides such as the Nextel polycrystalline Al_2O_3 fibers are not as attractive for most engine components, primarily because of their low creep strength at high temperatures (Yun and Goldsby, 1993). However, they may compete more effectively if they are less susceptible to strength degradation.

In summary, the future directions for the Ti-MMCs appear to be in the areas addressing actual turbine engine application. Thus, manufacturing technology for lower cost and improved reliability, the methodology needed for efficient design and accurate life prediction, and the accumulation of actual engine test experience should be emphasized. Improved MMC performance through matrix alloy development, environmental protection systems, and the use of alternate fibers are also logical choices. For the less mature superalloy matrix composites, laboratory-scale feasibility demonstrations are still required before engine applications can be seriously considered. The choices for fiber and the development of fiber coatings are currently limiting progress towards the goal of demonstrating mechanical properties which can compete with monolithic superalloys and intermetallics.

3.2. *Deformation and damage of MMC/IMCs*[†]

To fully realize the benefits offered by MMCs, *experimentally verified*, computationally efficient design and life-prediction methods must be developed for the advanced multi-phased materials of interest in advanced engine and propulsion systems. Consequently, these analysis tools must admit physically based, viscoplastic deformation

[†]Contributed by Steven M. Arnold and Michael G. Castelli.

life models and be compatible with the finite element method in order to accurately describe the complex thermomechanical load histories typical in the aerospace structures of interest. Furthermore, in order to assist both the structural analyst and the material scientist in developing and utilizing these materials, these tools must encompass the various levels of scale for composite analysis.

To respond to this difficult challenge, parallel approaches wherein the starting point is at the micro- and macroscale have been established at LeRC in deformation and damage modeling and experimental characterization and verification. Clearly, each approach has its realm of applicability, with micromechanics focusing primarily upon applications involving fabrication, material development and life assessment, and the primary usefulness of the macroscale approach[†] being in the design and analysis of structural components. The motivation for pursuing two parallel, yet not mutually exclusive approaches, is heightened by the fact that no one approach is clearly superior, relative to the primary goal of developing accurate, computationally efficient, and experimentally validated analysis tools. For example, the macroscale approach is clearly the most computationally efficient, yet its accuracy may suffer in comparison to its more computationally intense micro counterpart, particularly when highly localized, nonuniform behavior relative to the representative volume element (RVE) dominates.

Significant progress has been made over the past decade in the area of deformation and damage, with regard to experimental, theoretical and computational mechanics of composites (Arnold and Castelli, 1994). However, many issues still remain concerning experimental evaluation and "appropriate" material characterization for this class of materials. To date the vast majority of elevated-temperature experimental fatigue research has been conducted under uniaxial, load-controlled, tension–tension conditions on thin-plate coupons containing partially machined fibers. Great concerns remain within the experimental and modeling communities as to the effects of all of these variables and their relative impact on the data generated to date. Thus, the challenge and ultimate goal is to appropriately control and interpret the experimental evaluations so that accurate input can be provided to guide theoretical modeling efforts and verify their accuracy.

Numerous models both at the micro and macro scales have been proposed (see Arnold and Castelli, 1994, for a more thorough review). However, verification, particularly under thermomechanical multiaxial states of stress, and down selection of these various models is still needed. The dual approach at LeRC, wherein the analysis of structures is viewed both from the micromechanical and macromechanical standpoint, will continue. Ultimately the goal is to develop a hybrid approach for both deformation and damage that is both computationally efficient and accurate under general nonisothermal, multi-axial loadings. Consequently, one future trend will therefore be in the area of symbolic and parallel computations, so as to capitalize on the advances made in software design and computer architecture. Also, for multi-axial verification purposes, benchmark structural testing and analysis will be extremely important and pursued vigorously. It is our expectation that within the next decade accurate and computationally efficient design and analysis techniques will be developed and experimentally verified for a wide range of advanced composite systems with respect to high-temperature, time-dependent deformation and damage, thereby encouraging their assimilation into industry.

4. CERAMIC-MATRIX COMPOSITES

4.1. *Materials research*[‡]

Ceramics offer the potential to operate uncooled or with less cooling and at higher material temperatures than superalloys. This potential plus their low density and good resistance to oxidation make materials such as silicon nitride and silicon carbide extremely attractive. However, their brittle behavior and resultant sensitivity to small flaws that are

[†] The macroscale (continuum) approach is where the composite is considered as an isotropic material in its own right, with its own experimentally measurable properties.

[‡] Contributed by Stanley R. Levine.

either inherent in the as-produced material or which develop in service have precluded their reliable use in gas turbines. Recent progress, primarily in Japan and the United states, has demonstrated that fracture toughness, high temperature strength, and statistical reliability can be simultaneously improved. These improvements have resulted in improved functional reliability with durabilities on the order of 100s of hours in prototype automotive gas turbines and the ability to withstand major impact events. However, invariably something unanticipated occurs to cause catastrophic fracture. Thus, the question of the technical feasibility of ceramics for terrestrial engines remains open along with the question of economic viability vis-a-vis more conventional metal engines (Anon, 1993).

Aircraft gas turbine engines require an even higher degree of reliability. It is doubtful that monolithic or *in situ* toughened ceramics can achieve the required functional reliability levels in highly stressed rotating components due to the temporal nature of the flaw population. However, for small, low-stress, static components, they have proven viable (Levine and Herbell, 1992). Because unreinforced ceramics are subject to catastrophic fracture behavior and low reliability due to flaws, NASA Lewis has focused on fiber-reinforced ceramics for about the past 10 years. Our goal has primarily been to identify and develop fiber-reinforced ceramics with performance capabilities beyond those of superalloys in aircraft gas turbines. Therefore we have generally left to the industry, the pursuit of lower-temperature capability systems based on off-the-shelf fibers. We have primarily emphasized the development and characterization of advanced fibers, interphases and systems. Much progress has been made in materials, but many obstacles remain. These are discussed below.

The key to reliable, durable, strong, tough and affordable continuous-fiber-reinforced ceramics resides primarily with the reinforcements. The characteristics we seek are: high strength and stiffness, low density, matrix compatibility both chemically and with respect to thermal expansion match, small diameter for handleability, weaveability and optimum toughening, good thermal and microstructural stability, and, finally, affordable cost. Many of these attributes are also desired in ceramic fibers for reinforcement of metal and intermetallic-matrix composites (DiCarlo, 1991).

To support our interest in fiber development for ceramic-matrix composites (as well as metal- and intermetallic-matrix composites), we have invested considerable effort and resources in the development of fiber-characterization capabilities. Our facilities include equipment for measurement of fiber fast-fracture strength and elastic modulus at room to elevated temperature in air, vacuum and inert environments, and a laser-speckle strain-measurement system for elastic property measurements. The latter is under continued development to provide precision strain measurements for tensile fast fracture and creep. Also in place are systems for measurement of creep and stress rupture in air, vacuum and inert environments. The very simple bend-stress relaxation test (BSR) developed at LeRC has provided the industry with a simple, quick and readily implemented test for assessment of the relative creep resistance of fibers (Morscher and DiCarlo, 1992). Finally, in conjunction with other laboratories, we are contributing to the development of standardization of fiber test methods.

As a frame of reference for discussing fiber status, data will be limited to bend-stress relaxation comparisons. In the BSR test, a straight fiber is constrained to a uniform radius of curvature by tying it into a loop or placing it in a fixture (R_0). After high-temperature heat treatment, the constraint is removed and the radius of curvature is measured (R_A). If the fiber retains the radius of curvature of the constraint, it has fully relaxed (poor creep resistance). For this case, m, the bend-stress relaxation ratio, is 0 ($m = 1 - R_0/R_A = 0$). If it returns to its original straight shape, $R_A = \infty$, and no relaxation (or creep) has occurred ($m = 1 - R_0/R_A = 1$).

Many of the fiber characteristics discussed above for optimum fiber performance are best satisfied by stoichiometric silicon carbide. Several approaches to fabrication for silicon carbide fibers are showing promise for attaining good high-temperature stability and strength. The chemical vapor deposition approach of Textron Specialty Materials has yielded a variety of fiber microstructures and chemistries. The ability to tailor and control the process has yielded a 50 micron fiber with the best combined strength and creep resistance seen to date. Diameter reduction is still an issue. Carborundum has produced creep resistant α-SiC fibers by sintering of extruded green fiber. At this stage of development, the tensile strength is less than desired and the surface roughness and diameter are on the high side. Finally, Dow Corning has produced a near-stoichiometric SiC fiber by the polymeric precursor pyrolysis route. This fiber exhibits good creep resistance, high tensile strength and good handleability. One can conclude from the above that cost and the need for handleability, which is application driven, will be decisive factors in fiber selection. These four fibers along with commercially available fibers are compared in terms of bend-stress relaxation in Fig. 8 (DiCarlo, 1994).

A major concern with SiC fiber-reinforced ceramics is the oxidation resistance of the fiber and the fiber-matrix interface. One approach that can eliminate the interface oxidation issue is the use of oxide fibers and interphases in an oxide matrix. A number of textile-quality multi-filament oxide fibers based on either alumina or aluminosilicate compositions are commercially available. Since the creep of these fibers limits them to low use temperatures (<1100°C), they have not been the focus of our composites research. We have instead sought to identify fibers that have greater capability than single crystal sapphire. We are examining the potential of doped sapphire (Sayir *et al.*, 1993) and various eutectic compositions such as ZrO_2–Al_2O_3 (Farmer *et al.*, 1993) and YAG–Al_2O_3 for potential to offer greater toughness and better high-temperature strength retention and slow crack-growth resistance than sapphire. Exploratory research in the growth of these advanced fibers is being carried out by the laser-heated floating zone approach. Promising fibers are then transitioned to the commercial edge-defined film-fed growth process.

Interfaces with proper weak bonding and oxidative stability are also critical to the satisfactory mechanical performance of fiber-reinforced ceramics. The weak bonding requirement has been achieved with carbon coatings on the fibers in silicon carbide fiber-reinforced systems. However, the oxidation of the carbon interface and SiO_2 formation on the fibers (and matrix if SiC or Si_3N_4) results in bonding of the fibers to the matrix and a loss of strength. Surface coatings of the composite, dense matrices and SiC overlayers on the carbon-coated fibers can alleviate this problem, but do not represent a reliable long-term solution. Other approaches are being pursued at NASA LeRC including Ti–Si–C, BN and porous oxides. Oxide fiber-reinforced systems also require fiber coatings. For single crystal fibers, we have been examining porous oxides and highly anisotropic oxides. Application methods include sol gel, polymeric precursors, and CVD/CVI.

The composite systems being pursued at NASA LeRC can be classified by the type of reinforcement. With SiC reinforcements, we are investigating reaction-bonded silicon nitride (RBSN) and silicon carbide by silicon melt infiltration (Bhatt and Behrendt, 1992). Polymeric precursor fabrication approaches are also being pursued (Hurwitz, 1992). RBSN is attractive because the silicon nitridation process produces essentially no

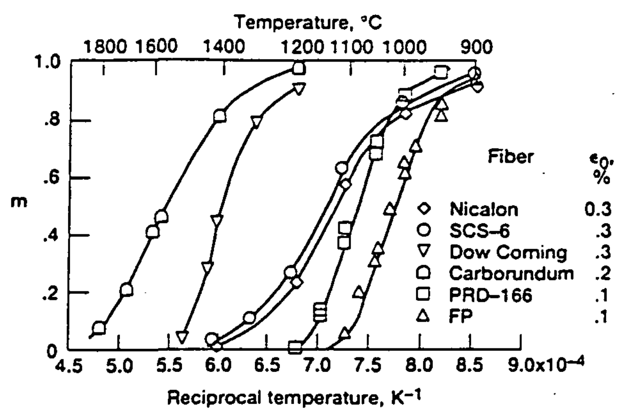


Fig. 8. One hour stress relaxation ratio, *m*, vs reciprocal temperature for several ceramic fibers (Tressler and DiCarlo, 1993).

dimensional change and thus the approach is strain compatible with fiber reinforcement. The polymeric precursor approach is attractive for its complex shape potential. Finally, the melt infiltration approach is attractive because it can yield fully dense matrices.

Our melt infiltration (MI) approach, called reaction forming, is carried out by forming a carbon precursor matrix of controlled porosity and pore size by pyrolysis of a foamed polymer. This allows thorough and uniform silicon melt infiltration and conversion to a silicon carbide plus residual silicon matrix whose microstructure is controlled by the precursor network. Alloying of the silicon with niobium or molybdenum allows the introduction of a third phase for tailoring of toughness, strength and thermal expansion. This capability is illustrated by the photomicrograph in Fig. 9 (Singh *et al.*, 1994). Composites can be produced by resin transfer molding (RTM) or by ply lay-up. However, this process can be combined with chemical vapor infiltration to yield a hybrid processing approach. The basis for this approach is that all SiC fiber-reinforced composites will require an interface coating on the fibers. The most reliable and cost-effective method for placement of this coating would be at the woven preform stage using chemical vapor infiltration (CVI). Furthermore, this coating can be protected by a SiC overlayer coating and the preform rigidized by some additional CVI SiC. From this point, densification can be carried out rapidly and economically by the reaction-formed silicon carbide process. The carbon precursor can be placed by resin transfer molding (RTM), pyrolyzed and converted to SiC by Si melt infiltration (MI). This approach has been dubbed CRM for *C*VI, *R*TM, *M*I.

In the oxide matrix arena, we have been investigating the celsian family of glass-ceramic matrices. Our starting point was barium-aluminosilicate (BaO–Al_2O_3–$2SiO_2$) or BAS. This glass ceramic offers higher temperature capability than other glass ceramics commonly reported as composite matrix materials (e.g. LAS, MAS, BMAS). Bend strength for SCS-6- reinforced BAS is shown in Fig. 10 in comparison to the unreinforced matrix (Bansal, 1994). Strontium substitution for barium in total or in part is being investigated for improved processability, and small-diameter fiber reinforcements are also being investigated. In addition to glass ceramics, we are looking at crystalline matrices such as mullite and alumina combined with either SiC or single-crystal oxide reinforcements.

In summary, the identification of strong, stable and weavable fibers and durable interfaces continue to be very high priority areas for CMC research. Advanced fibers and interfaces are being incorporated into microcomposites and, as sufficient fiber quantities become available, into coupons for assessment of mechanical and environmental durability behavior. The more mature and promising systems are being advanced to rig and engine tests as quickly as possible.

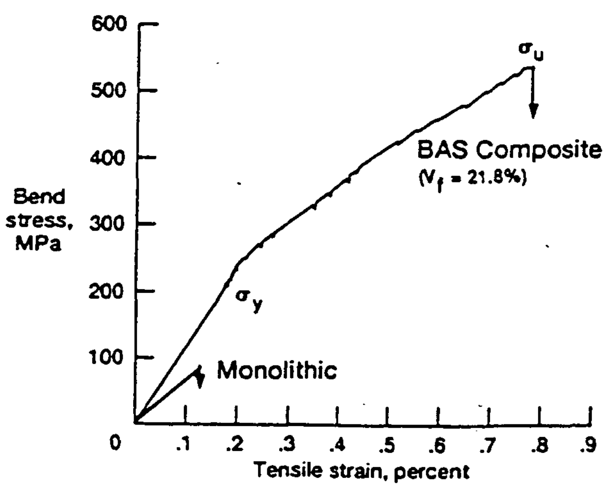


Fig. 10. Bend strength of SCS-6 fiber-reinforced barium-aluminosilicate shows greater strength and strain capability than the monolithic material (Bansal, 1994).

4.2. *Current trends in CMC component analysis*[†]

From an aerospace design engineer's perspective, ceramic composites offer significant potential for raising the thrust/weight ratio and reducing NO_x emissions of gas turbine engines. Considering that these materials will be produced from abundant nonstrategic materials, it is not surprising that research has focused on improving ceramic material properties through processing, as well as establishing protocols for sound design methodology. In particular, continuous ceramic fiber composites exhibit an increase in work of fracture, which allows for "graceful" rather than catastrophic failure. When loaded in the fiber direction, these composites retain substantial strength capacity beyond the initiation of transverse matrix cracking despite the fact that neither of their constituents would exhibit such behavior if tested alone. Indeed, first matrix cracking consistently occurs at strains greater than that in the monolithic matrix material. As additional load is applied beyond first matrix cracking, the matrix tends to break in a series of cracks bridged by the ceramic fibers. Thus any additional load is borne increasingly by the fibers until the ultimate strength of the composite is reached. For most applications the design failure stress will be taken to coincide with the first matrix cracking stress. Matrix cracking usually indicates a loss of component integrity since this phenomenon allows high-temperature oxidation of the interface and fiber, which leads to the strength loss of current composites.

The analysis and design of components fabricated from ceramic composite materials require a departure from the usual deterministic design philosophy (i.e. the factor of safety approach) prevalent in the analysis of metallic structural components, which are more tolerant of flaws and material imperfections. Under applied load, large stress concentrations occur at macroscopic as well as at microscopic flaws, which are unavoidably present in the composite as a result of processing or in-service environmental factors. The observed scatter in component strength is caused by various failure mechanisms, and their corresponding severity leads to composite fracture when the damage-driving force or the effective energy release rate reaches a critical value. This scatter is evident in Fig. 11, where the uniaxial failure data for an oxide–oxide ceramic composite are depicted (Ye, 1994). The data represent the first matrix cracking stress associated with the fiber direction of an alumina matrix reinforced with polycrystalline alumina fibers. Note that the Weibull modulus estimated from this data is 3.68. This value is an indication that significant scatter in composite microcracking strength is present. Observe that the largest stress value in this data set represents over a 330% increase from the lowest level. We should also note that a number of deterministic micromechanical fracture theories exist in the literature that predict a composite's first matrix cracking strength as a function of its constituents. Since all are based on assumed idealistic

[†] Contributed by Stephen Duffy.

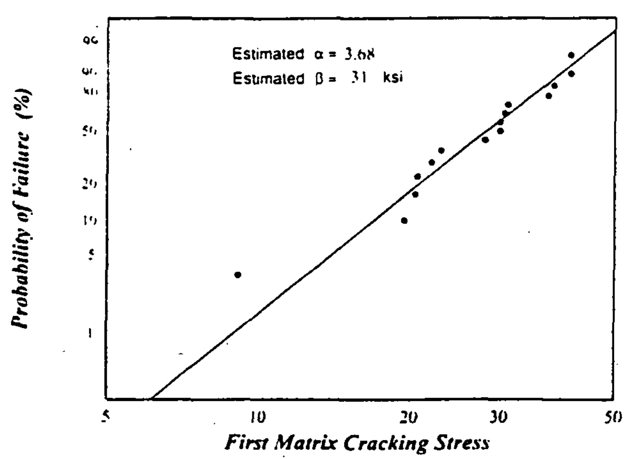


Fig. 11. Uniaxial first matrix cracking stress data for an alumina fiber-reinforced alumina composite.

microstructures, they are typically unable to predict the unavoidable strength variation in current-generation composite materials. In addition, most ceramics exhibit decreasing bulk strength with increasing component volume (the so-called size effect). Since failure is governed by the scatter in strength (ultimate or microcrack yield), statistical design approaches must be employed.

Utilizing structural reliability methods provides a more general accounting of the entire spectrum of values that strength parameters may exhibit. However, the reliability approach demands that the design engineer must tolerate a finite risk of unacceptable performance. This risk of unacceptable performance is identified as a component's probability of failure. The primary concern of the engineer is minimizing this risk in an economical manner. Most quantities that are utilized in engineering designs have, to a greater or lesser extent, some level of uncertainty. This means that if reliability methods are utilized, appropriate analytical tools needed to quantify uncertainty must be readily available. A number of tools and design aids for dealing with uncertainty in a rational fashion have been developed here at NASA Lewis Research Center. These tools include reliability models and computer software that have been tailored to specific composite systems. The reader is directed to the work by Thomas and Wetherhold (1991), Duffy and Arnold (1990), Duffy and Manderscheid (1990) and Duffy *et al.* (1993), regarding the development of reliability models. A number of these reliability models have been incorporated into public-domain computer algorithms such as the T/CARES (Toughened Ceramics Analysis and Reliability Evaluation of Structures) and C/CARES (Composite Ceramics Analysis and Reliability Evaluation of Structures). These computer algorithms are coupled to an assortment of commercially available general-purpose finite element programs. The algorithms yield quasi-static component reliabilities of structures fabricated from ceramic composites; however, work is underway to formulate time-dependent algorithms. Current thought focuses on incorporating the principles of continuum damage mechanics in a similar manner outlined by Duffy and Gyekenyesi (1989).

Focusing attention on the C/CARES algorithm, a noninteractive reliability model has been incorporated where individual uniaxial plies are treated as two-dimensional structures. Each ply (which is discretized in the analysis) is assumed to have five basic strengths or failure modes. The assumption is made that failure is governed by the strength of the weakest link. In essence the component is treated as a series system, and the component probability of failure is evaluated accordingly. Admittedly, the weakest-link concept is a somewhat conservative approach for composites where micro-redundancies exist in certain directions due to parallel arrangements of fibers. However, a macro-level approach to strength measurements should capture this behavior through enhanced distribution parameters. This distinguishes the T/CARES and C/CARES codes from other reliability software where the probability of point failure is usually evaluated. Treating a discretized component as a system allows the design engineer to evaluate size effects, which is not possible when the probability of point failure is evaluated. In addition, the CARES family of software includes parameter-estimation modules that allow the design engineer to evaluate the strength-distribution parameters from failure data. It is assumed that failure strengths can be characterized by either a two- or three-parameter Weibull distribution.

Recent progress in processing ceramic composites has not been matched by mechanical testing efforts. This type of data supports the creation of a complete design data base for a given material. In addition, there is a definite need for experiments that support the development of reliability models. Initially this effort should include experiments that test fundamental concepts (e.g. quantifying size effect in the fiber direction) within the framework of current stochastic models. For example, probing experiments should be conducted along various biaxial load paths to establish level surfaces of reliability in a particular two-dimensional stress space (similar to probing yield surfaces in metals). Concepts such as the maximum stress response which is often assumed in the noninteractive reliability models could be assessed. After establishing a theoretical framework, characterization tests should be conducted to provide the functional dependence of model parameters with respect to temperature. Finally, data from structural tests that are multiaxial (and possibly nonisothermal) could be used to challenge the predictive capabilities of models through comparison to benchmark response data. It cannot be overemphasized that this kind of testing supports design and analysis of components.

5. STRUCTURAL MECHANICS[†]

Achieving the full benefits of composites for aerospace-propulsion and power-system (engine) applications ultimately requires the availability of credible and efficient computer-based tools for component analysis and design. As implied earlier, tools are required which can account for both micromechanical and macromechanical factors affecting critical composite structural performance requirements. Some recent efforts to develop such tools, for a variety of composite materials and structural concepts, are briefly described below.

5.1. *Tools for high-temperature composites*

An important factor affecting the behavior of a CMC is the condition of the interface (or interphase) between fiber and matrix. A distinct interphase can exist as an intentionally applied fiber coating, or can arise due to chemical reaction that occurs between the fiber and matrix during composite fabrication and/or during service at elevated temperatures. The stiffness, strength and thickness of an interphase will influence the overall thermomechanical behavior of a CMC.

One computer-based tool under development at NASA Lewis Research Center, known as CEMCAN (for CEramic Matrix Composite ANalyzer), has recently been used to investigate interface (interphase) effects on CMC behavior. CEMCAN implements a unit cell or representative volume element (RVE) approach with a novel fiber substructuring technique. In this technique the fiber is substructured into multiple layers and the micromechanics equations are formulated at the layer level. The RVE also incorporates a distinct fiber/matrix interphase constituent.

In recent applications of CEMCAN, a unidirectional SiC/RBSN composite (silicon carbide SCS-6 fibers in reaction-bonded silicon nitride matrix) was analyzed for both strong and weak fiber/matrix bond conditions (Mital *et al.*, 1993a). In a strong bond condition, the thermoelastic properties of the distinct interphase constituent are taken to be the same as the matrix (upper bound), whereas in the case of a weak bond, the normal and shear elastic moduli of the interphase are reduced to negligible values (lower bound).

The predicted values of composite effective properties are compared to experimentally measured values, wherever available, and the properties of the interphase are calibrated. The variation of composite properties can also be predicted for varying extent of debond around the fiber circumference or interfacial damage through-the-thickness of the composite.

Results indicate that longitudinal composite properties are rather insensitive to bond conditions, while transverse composite properties are influenced significantly by the bond conditions. Moreover, the comparisons between CEMCAN predictions and experimentally measured values for a SiC/RBSN composite show good agreement as illustrated, for example, in Fig. 12. If the interfacial debonding/damage is limited to a few plies, the degradation in the composite properties is minimal and perhaps difficult to detect by conventional experimental measurements.

The primary advantage of a tool such as CEMCAN is that it provides a simplified, but flexible, capability to represent complex factors such as varying degrees of interfacial bond around the fiber circumference or through-the-thickness, local matrix cracking and fiber breaks (Mital *et al.*, 1993b), different fiber shapes, etc., and the integrated effect of all these aspects on the composite effective properties and thermomechanical behavior. The fiber substructuring technique also permits more accurate (in a piece-wise sense) resolution of local stress distributions in the composite constituents (fiber, interphase and matrix).

[†] Contributed by Dale A. Hopkins.

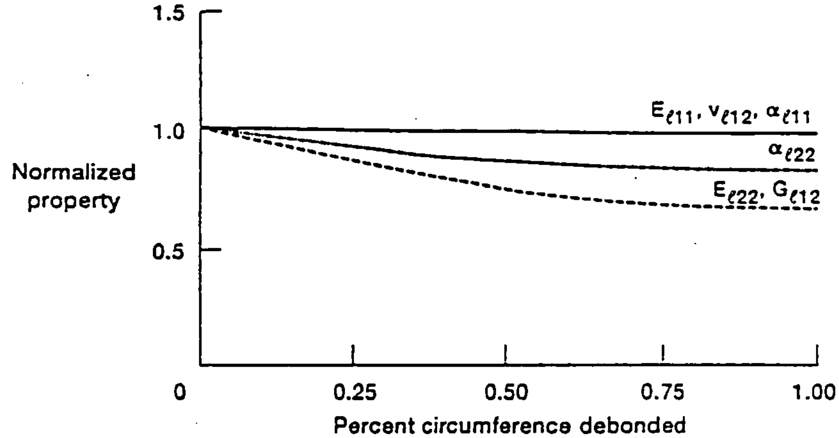


Fig. 12. Unit-cell model and effective elastic properties for [0] SiC/RBSN composite with simulated effect of incomplete fiber/matrix bond.

Another computer-based tool under development at NASA Lewis Research Center, known as BEST—CMS (for *B*oundary *E*lement *S*olution *T*echnique—*C*omposite *M*odeling *S*ystem), has also recently been used to investigate interface effects on CMC behavior. As the name implies, BEST—CMS employs an innovative discrete boundary element methodology and provides sophisticated capabilities for modeling arbitrary fiber architectures, complex fiber/matrix interface conditions, and complex material constitutive behaviors.

The above notwithstanding, a major advantage of BEST—CMS lies in its extremely simple discretization requirements. Specifically, a BEST—CMS model of the composite entails discretizing only the exterior surface of the matrix (with 2-D surface elements) and the centerlines of fibers (with 1-D line elements). The fiber/matrix interface condition is specified merely by entering the type (perfect bond, linear spring, nonlinear spring, or frictional sliding) and the corresponding spring parameters and/or friction coefficient. No explicit discretization of the fiber/matrix interface is required, as the interface behavior is incorporated through the underlying boundary integral equation formulation.

The modeling simplicity advantages of BEST—CMS are more apparent when contrasted to what would be required to create an equivalent finite element model. In the latter case, the entire volume of fibers and matrix must be discretized with 3-D solid elements, and the complex fiber/matrix interfaces must be explicitly modeled using special techniques such as gap elements.

The benefits alluded to above are illustrated in Fig. 13 which shows a sample BEST—CMS model of a unidirectional laminate and computed stress-strain behavior resulting from a linear stress analysis (Goldberg and Hopkins, 1993). The stress-strain results are for a $[90]_2$ SiC/RBSN composite, with fiber/matrix interface conditions specified to simulate both perfect and imperfect bonding. The computed results are compared to experimentally observed behavior, with the simulated imperfect bond case showing better agreement.

In summary, two alternative computer-based tools and their use for micromechanical analyses of CMCs have been described. The capabilities of these tools to model complex factors such as fiber/matrix interface conditions have been demonstrated, and some degree of credibility has been established through comparisons with experimental observations. Whereas the previous focus has been on straight-fiber laminated composites, future emphasis will shift toward more complex composite architectures such as weaves and braids. Indeed, woven and braided PMCs and CMCs are already being pursued for potential engine applications. Accordingly, more sophisticated tools will be needed to enable credible and efficient engine component analysis and design procedures. The BEST—CMS tool, for example, shows particularly good promise in this domain.

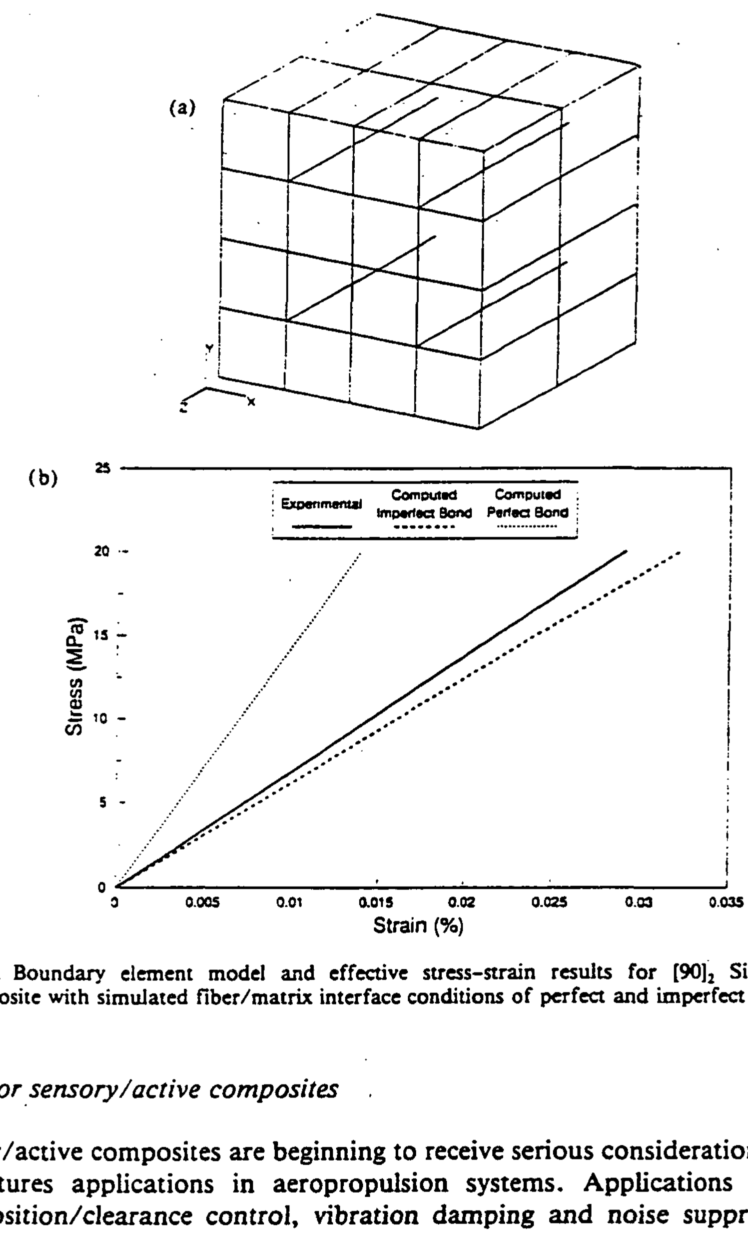


Fig. 13. Boundary element model and effective stress–strain results for $[90]_2$ SiC/RBSN composite with simulated fiber/matrix interface conditions of perfect and imperfect bond.

5.2. *Tools for sensory/active composites*

Sensory/active composites are beginning to receive serious consideration for various smart structures applications in aeropropulsion systems. Applications include, for example, position/clearance control, vibration damping and noise suppression. Very recent progress has occurred in the development of composite mechanics and structural analysis models which is leading toward computer-based tools for the analysis and design of smart engine components.

A unified composite mechanics theory was developed with the capability to model laminated composite structures with embedded piezoelectric layers for both sensory and active modes of behavior (Heyliger and Saravanos, 1993). Using a discrete-layer representation for both displacement and electric potential fields, the theory can accurately model global as well as local electromechanical response. The inclusion of electric potential into the state variables allows representation of general electromechanical boundary conditions and facilitates integration with controller models or other electronic components. Moreover, the formulation includes all energy contributions from elastic, piezoelectric and dielectric components.

The formulations for static and dynamic response of smart composite beam and plate structures with embedded sensors and actuators have also been completed, and specialty finite elements were developed for this purpose. Evaluations have demonstrated the capability of these formulations to represent either sensory (see Fig. 14) or active structures, and to model the complicated stress–stain fields including interactions between passive and active layers (see Fig. 15), interfacial phenomena between sensors and

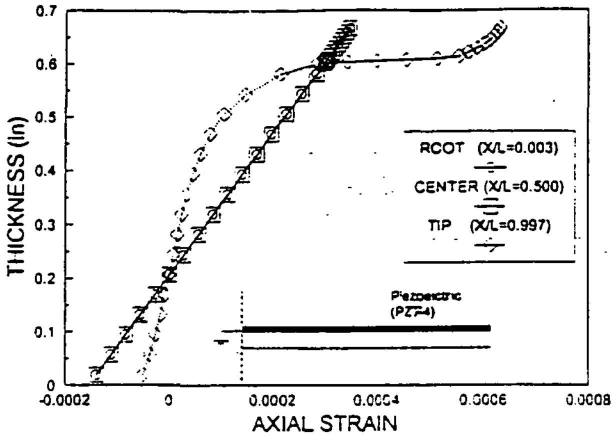


Fig. 14. Through-the-thickness variations of axial strain in active T300/934 composite beam with single piezoceramic layer.

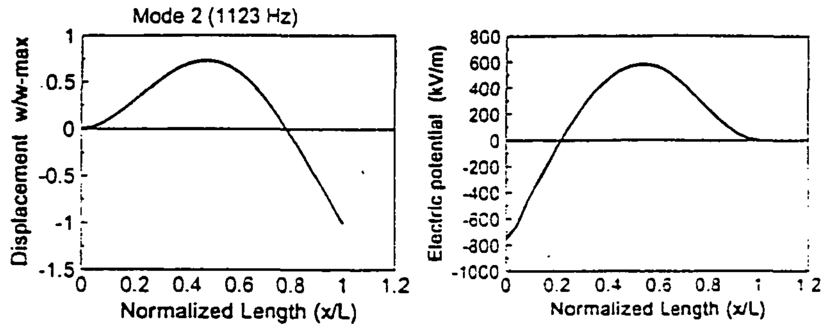


Fig. 15. Typical mode shape and associated electric potential in top surface of sensory beam (2nd bending mode).

composite plies, and critical damage modes in the material. Furthermore, the capability to predict dynamic characteristics under various electric circuit configurations has been demonstrated.

The analytical foundations have also been developed to enable the application of sensory composite structures with delamination-failure detection capabilities by monitoring changes in their dynamic characteristics (Saravanos, 1993). Such non-destructive, real-time health-monitoring capabilities may dramatically improve the reliability of aerospace structural composites. In this work admissible composite mechanics were formulated enabling representations of the effects of delamination cracks and disbonds on the laminate properties such as stiffness, damping, inertia, stresses, etc. An exact analytical procedure was further developed for the prediction of natural frequencies, mode shapes and modal damping in composite beams with an interlaminar delamination.

Evaluations for various cantilever beams with a central delamination have been completed. Correlations with limited reported experimental results show excellent agreement (see Fig. 16). The results indicate that natural frequencies are rather insensitive to small delamination cracks. On the other hand, modal damping seems to be a superior indicator of delamination damage, yet the effects of delamination on damping may vary based on crack size, laminate configuration and mode order. Thus, the combination in changes in both damping and natural frequencies seems to provide a damage signature which may lead to the detection of delamination cracks.

Overall, the mechanics have provided valuable insight into the problem, have facilitated the interpretation of experimental results, and have demonstrated the feasibility of smart composite structures with health-monitoring capabilities. More importantly, the mechanics models have provided the missing link which will enable real-time, in-service detection of delamination presence, size and location from changes in the dynamic signature of the composite structure.

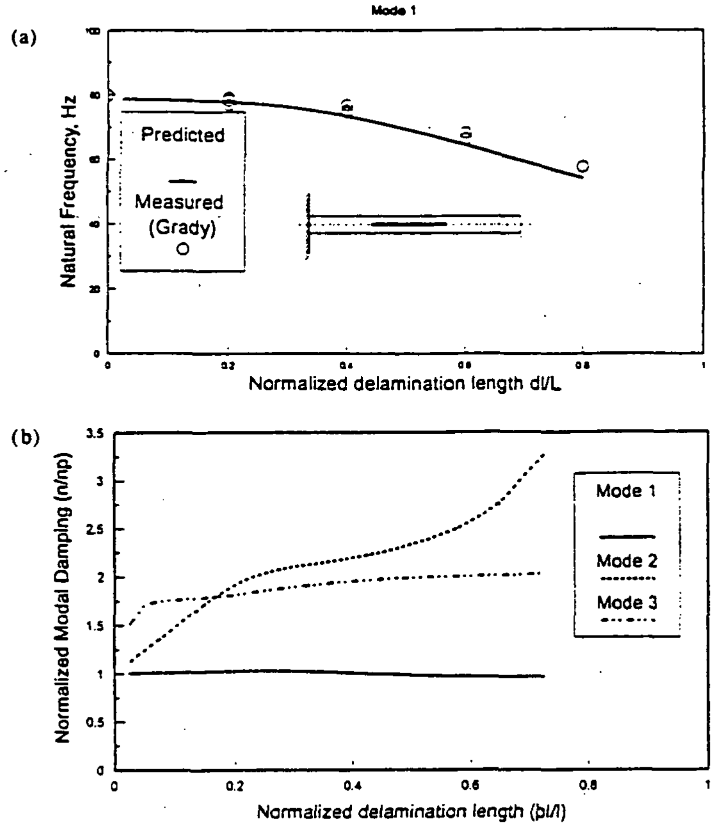


Fig. 16. Effect of delamination size on: (a) fundamental natural frequency of $[0/90]_{2S}$ T300/934; (b) and modal damping of a $[45/-45/90/0]_S$ T300/934 beam.

In summary, considerable progress has been made to develop the fundamental mechanics and structural analysis models necessary to confidently predict the response of sensory/active smart composites. Future efforts will examine other issues, such as digital control systems, power requirements, operating limits, etc., necessary to establish the practical feasibility of smart composite structures in engines.

6. NONDESTRUCTIVE EVALUATION (NDE)†

Composites for advanced high-temperature, high-efficiency engines pose new and special challenges for NDE. The engine components will consist of a variety of polymeric-, intermetallic- and ceramic-matrix composite structures. The complex nature of these structures creates strong incentives for advanced nondestructive interrogation and evaluation methods. NDE must range from detection of individual flaws to global imaging of fiber architecture and probabilistic assessment of diffuse flaw populations (Vary, 1992).

At NASA Lewis Research Center, we approach structural composites from the viewpoint that the detection and resolution of individual *micro*-flaws may be unnecessary. This does not mean that individual *macro*-flaws such as delaminations, cracks and similar discontinuities may be ignored. However, it should be recognized that composites may contain a profusion of minute defects that have no discernable effect on reliability or performance unless they are in close proximity and interact massively or encourage degradation in service environments. Then, the challenge is to characterize the collective effect of several kinds of subcritical flaws on mechanical integrity and strength. This is in addition to the detection of overt, dominant defects or global aberrations that would have adverse effects on structural integrity.

†Contributed by Alex Vary.

Our view is that NDE methods should be applied concurrently in engineering design, process modeling and structural life analysis. This is in addition to NDE (a) during raw material processing to assure quality, (b) in early stages of component fabrication to screen out defective parts, (c) after fabrication to verify structural integrity, and (d) following service to assess thermomechanical degradation and residual life. NDE methods also provide powerful tools for materials characterization and, as indicated below, we are exploiting these methods in materials testing research to help develop fracture- and life-prediction codes.

The LeRC Structural Integrity Branch investigates and develops methods for nondestructive materials interrogation and flaw characterization. The Branch concentrates on radiographic and ultrasonic techniques including micro-focus radiography, computed tomography, scanning acoustic microscopy, and laser ultrasonics. LeRC researchers pioneered the acousto-ultrasonic (AU) technique which is a practical, sensitive NDE method for assessing variations and degradations of mechanical properties of composites (Vary, 1990). A **Standard Guide for Acousto-ultrasonic Assessment of Composites, Laminates and Bonded Joints** was adopted by ASTM's Committee E-7 on Nondestructive Testing (Anon, 1993).

LeRC researchers are currently exploring *in situ* NDE monitoring of damage accumulation processes in ceramic-matrix composites during mechanical destructive testing. This work combines conventional load frame instrumentation with nondestructive interrogation methods. The *in situ* NDE methods involve adaptations of radiographic, acoustic emission, acousto-ultrasonic, thermographic, and laser imaging techniques. The idea is to apply NDE methods during destructive testing to better understand materials response and to validate fracture prediction and damage accumulation models. This can enhance the various inspection opportunities mentioned previously and the reliability assessment of advanced composite structures before and following service.

An example of *in situ* NDE is our use of radiographic images to determine matrix crack spacing from which one can calculate interfacial shear strengths on the basis of the Aveston–Cooper–Kelly (ACK) theory (Chulya *et al.*, 1991). The *in situ* X-ray method is superior to the conventional optical method for determining crack spacing. The X-ray method provides full-field images of matrix cracking through the entire volume of the gauge section. This NDE method is preferable for materials characterization in that it does not require unloading and removing specimens which would result in crack closure and errors in determining crack spacing.

We apply acoustic emission (AE) and acousto-ultrasonic (AU) methods *in situ* during tensile loading of fiber-reinforced ceramics, i.e. ceramic-matrix composites (CMCs), to identify and discriminate among various failure mechanisms. The objective is to validate "first fracture" and life-prediction models. Our work has identified fracture mechanisms via AE and AU parameters. For example, AU parameters provide relations between stress levels and the onset and saturation of matrix cracking (Tiwari and Hennecke, 1993). AU was is also useful in determining an "effective ultrasonic modulus". We found that this modulus provides a good measure of interfacial shear strength and correlates with modulus values determined from tensile tests. These results confirm that AU can provide a viable approach to nondestructive monitoring of mechanical property changes in composites.

LeRC has installed facilities for experimental study of composite fracture, mechanical response and durability under extreme environmental conditions. The strength, stiffness, toughness and fatigue crack-growth parameters can be evaluated at temperatures to 3000°F (1650°C) in inert, air and other gaseous environments. These test facilities meet the challenges associated with establishing mechanical test methods, sample specifications, and characterization of high-temperature composites over a wide range of thermomechanical conditions. Probabilistic models and algorithms are being developed for sensitivity analyses needed for identifying and predicting the effects of defects and constitutive parameters on the behavior of composites. We expect these latter efforts to provide foundations for guiding NDE for reliability assessment and life prediction.

7. CONCLUDING REMARKS

Historically, most new materials have established viable markets and found commercial success through a linear product-development cycle. In the past the materials scientist would develop a new material system, prototypical components were then fabricated and tested, data bases would be established, and design methodologies were developed in a sequential process. This linear product-development approach was adequate during the cold-war era when large research and development budgets spawned a number of successful high-technology material systems (e.g. smart materials, the utilization of composite materials in the air frames of jet fighters, etc.). However, as American industry continues the struggle to constantly reinvent itself in the post-cold-war era, artifacts such as the linear material-development cycle are being discarded. The current political climate, reduced budgets, and the need to develop dual-use technology all demand that economic issues (and not national defense needs) will dominate the direction of materials research and development. The materials community, which includes material scientists and product design engineers (both at the national research labs and within American industry), must adopt new integrated product-development teams that utilize an assortment of multi-disciplinary skills. In addition, these integrated product-development teams must involve end-users early in the development cycle to ensure economic viability. A primary goal of the integrated product-development teams must be a reduction in the material development cycle. The competitiveness of American material suppliers and their product end-users demands that the cycle for product development be shortened. If a reduction in time-to-market is achieved, the direct results are more American jobs and an improved economic position for American industries in today's global market.

A reduction in the development cycle requires that the concepts of concurrent engineering be embraced. Moreover, to establish a concurrent engineering infrastructure for composites, design guidelines must be established early through codes and standards organizations such as ASTM and ASME. Unless there is a tremendous cost saving or system enhancement (e.g. the NO_x emission reduction in jet engines mentioned previously), product engineers will not utilize a new material until they are comfortable knowing that an appropriate design practice has been codified. The reader need only study the commercialization (or lack thereof) of polymer-matrix composites and carbon–carbon composites to find the evidence to support this last statement.

At NASA Lewis, we have adopted this philosophy in the execution of technology programs for an advanced subsonic transport (AST) and high-speed civil transport (HSCT) and will continue to apply it in developing new initiatives. These programs are being carried out by integrated teams of industry, university, and NASA researchers focused on the needs of the end-use customer. In addition, as the aerospace industry continues to down-size, the longer-term and more research-oriented aspects of the business are being cut to the bone. As a result, there is a strong dependence on NASA to support longer-term research and technology-base efforts as well as near-term focused research.

REFERENCES

Anon (1992). *Proc. Annual Automotive Technology Development Contractors Coordination Meeting*, p.265. Society of Automotive Engines.

Anon (1993). Standard guide for acousto-ultrasonic assessment of composites, laminates, and bonded joints. *Annual Book of ASTM Standards*, 03.03, *Nondestructive Testing*, pp. 760–767. ASTM, Philadelphia, PA.

Arnold, S. M. and Castelli, M. G. (1994). Continuum-based theoretical and experimental studies in deformation and damage of MMCs at NASA–Lewis: progress and trends. *Compos. Engng* 4(8), 811–828 (this issue).

Bansal, N. P. (1994). Method of producing a cermic fiber-reinforced glass-cermic matrix composite. U. S. Patent 5,281,559.

Bhatt, R. T. and Behrendt, D. R. (1992). Reaction-bonded Si_3N_4 and SiC matrix composites. Chapter 6 of *Flight-vehicle Materials, Structures, and Dynamics—Assessment and Future Directions*, Vol. 3, *Ceramics and Ceramic-matrix Composites*, pp. 101–111. ASME, New York.

Bowles, K. J. (1990). Thermo-oxidative stability studies of PMR-15 polymer matrix composites reinforced with various continuous fibers. *SAMPE Q.* **21**, 6013.

Bowman, R. R. (1992). Influence of interfacial characteristics on the mechanical properties of continuous fiber reinforced NiAR composites. In *MRS Symp. Proc.* no. 273, *Intermetallic Matrix Composites* Vol. II (Edited by D. B. Miracle, D. L. Anton and J. A. Graves), pp. 145-156. Materials Research Society.

Brindley, P. K. and Draper, S. L. (1993). Failure mechanisms of 0° and 90° SiC/Ti-24Al-11Nb composites under various loading conditions. In *Structural Intermetallics* (Edited by R. Darolia, J. J. Lewandowski, C. T. Liu, P. L. Martin, D. B. Miracle and M. V. Nathal), pp. 727-737. TMS, Warrandale, PA.

Castelli, M. G. (1993). Thermomechanical and isothermal fatigue behavior of a $[90]_8$ titanium matrix composite. *Proc. 8th Tech. Conf. on Composite Materials, Mechanics, and Processing*, pp. 884–892. American Society for Composites.

Chuang, K. C., Kinder, J. D., Hull, D. L. and Youngs, W. Y. (1994). Rodlike polyimides. *Polymer Preprints* 35(1), 341-342.

Chuang, K. C., Vannucci, R. D. and Moore, B. W. (1992). Effects of a noncoplanar biphenyldiamine on the processing and properties of addition polyimides. *Polymer Preprints* 33(1), 435–436.

Chulya, A., Gyekenyesi, J. P. and Bhatt, R. T. (1991). Mechanical behavior of fiber reinforced SiC/RBSN ceramic matrix composites: Theory and experiment. *ASME Pamphlet 91-GT-209*. ASME, New York (also NASA TM 103688).

Delvigs, P., Klopotek, D. L. and Cavano, P. J. (1994). Graphite fiber/polyimide composites with improved processability. In *MRS Symp. Proc.* no. 305, *High Performance Polymers and Polymer Matrix Composites* (Edited by R. K. Eby, R. C. Evers, M. A. Meador and D. Wilson), pp. 33-39. Materials Research Society.

DiCarlo, J. A. (1991). High temperature structural fibers—status and needs. NASA TM 105174.

Dicarlo, J. A. (1994). Creep limitations of current polycrystalline ceramic fibers, composites science and technology. *Compos. Sci. Technol.* **51**, 213-222.

Draper, S. L. and Locci, I. E. (1994). Al_2O_3 fiber strength degradation in metal and intermetallic matrix composites. *J. Mater. Res.*, in press.

Duffy, S. F. and Arnold, S. M. (1990). Noninteractive macroscopic statistical failure theory for whisker reinforced ceramic composites. *J. Compos. Mater.* **24**, 293-308.

Duffy, S. F. and Gyekenyesi, J. P. (1989). Time-dependent reliability model incorporating continuum damage mechanics for high-temperature ceramics. NASA TM-102046.

Duffy, S. F. and Manderscheid, J. A. (1990). Noninteractive macroscopic reliability model for ceramic matrix composites with orthotropic material symmetry. *Trans. ASME, J. Engng for Gas Turbines and Power* **112**, 507-511 (also published as NASA TM-101414).

Duffy, S. F., Palko, J. L. and Gyekenyesi, J. P. (1993). Structural reliability analysis of laminated CMC components. *Trans. ASME, J. Engng for Gas Turbines and Power* **115**, 103-108 (also published as NASA TM-103685).

Farmer, S. C., Sayir, A. and Dickerson, P. O. (1993). Mechanical and microstructural characterization of directionally-solidified alumina-zirconia eutectic fibers. *Proc. Symp.* In Situ *Composites, Science & Technology*, pp. 167-182. TMS, Warrendale, PA.

Gabb, T. P., J. Gayda, Bartolotta, P. A. and Castelli, M. G. (1991). A review of thermomechanical fatigue damage mechanisms in two titanium and titanium aluminide matrix composites. *Int. J. Fatigue* **14**(1), 413–422.

Gayda, T. P. and Gabb, J. (1992). Isothermal fatigue behavior of a $[90°]_8$ SiC/Ti-15-3 composite at 436°C. *Int. J. Fatigue* **14**, 14–20.

Goldberg, R. K. and Hopkins, D. A. (1993). Micromechanical modeling of laminated composites with interfaces and woven composites using the boundary element method. NASA TM 106280.

Harding, D. R. and Sutter, J. K. (1993). Oxidation barrier coatings for high temperature PMCs. NASA CP-19117, 17-1-17-8.

Heyliger, P. R. *.noand Saravanos, D. A. (1993). On discrete-layer mechanics for health monitoring applications in smart composite structures. In Adaptive Structures and Material Systems* (Edited by G. P. Carman and E. Garcia), ASME Winter Annual Meeting, AD-Vol. 35, pp. 303-312. ASME, New York.

Hurwitz, F. I. (1992). Polymeric precursors for fibers and matrices. Chapter 4 of *Flight-vehicle Materials, Structures, and Dynamics—Assessment and Future Directions*, Vol. 3, *Ceramics and Ceramic-matrix Composites*, pp. 59-77. ASME, New York.

Kandebo, S. W. (1992). Allison tests variable cycle fighter/attack core engine. *Av. week and Space Tech.* **24 Feb.**, 130–131.

Larsen, J. M., Revelos W. C. and Gambone, M. L. (1992). Possibilities and pitfalls in aerospace applications of titanium matrix composites. In *MRS Symp. Proc.* no. 273, *Intermetallic Matrix Composites* Vol. II (Edited by D. B. Miracle, D. L. Anton and J. A. Graves), pp. 3-16. Materials Research Society.

Lerch, B. A. (1990). Fatigue behavior of SiC/Ti-15-3 laminates. NASA CP-10051, 35.1-35.9.

Levine, S. R. and Herbell, T. P. (1992). Aerospace applications. In *Engineered Materials Handbook*, Vol. 4, *Ceramic and Glasses*, pp. 1003-1006. ASM, Metals Park, OH.

MacKay, R. A., Brindley, P. K. and Froes, F. H. (1991). Continuous fiber-reinforced titanium aluminide composite. *J. Metals* 43(5), 23-29.

MacKay, R. A., Draper, S. L., Ritter and Siemers, P. A. (1994). A comparison of the mechanical properties and micro structures of intermetallic matrix composites fabricated by two different methods. *Met. Trans.*, in press.

Meador, M. A., Cavano, P. J. and Malarik, D. C. (1990). High temperature polymer matrix composites for extreme environments. In *Structural Composites: Design and Processing Technologies, Proc. 6th Annual ASM/ESD Advanced Composites Conf.*, pp. 529-539. ASM, Metals Park, OH.

Miller, L. M. and Gulino, D. A. (1994). Aluminum oxide films for thermo-oxidative protection of polyimide-based composites. In *MRS Symp. Proc.* no. 305, *High Performance Polymers and Polymer Matrix Composites* (Edited by R. K. Eby, R. C. Evers, M. A. Meador and D. Wilson), pp. 191-196. Materials Research Society.

Mital, S. K., Murthy, P. L. N. and Chamis, C. C. (1993a). Ceramic matrix composites properties/microstresses with complete and partial interphase bond. NASA TM 106136.

Mital, S. K., Murthy, P. L. N. and Chamis, C. C. (1993b). Ceramic matrix composites: effect of local damage on composite response. NASA CP 19117, Vol. III, pp. 76.1–76.11.

Morscher, G. N. and DiCarlo, J. A. (1992). A simple test for thermomechanical evaluation of ceramic fibers. *J. Am. Ceramic Soc.* **75**(1), 136–140.

Pickens, J. W. Smith, J. W. and Lerch B. W. (1993). Relationships among processing surface structure, and properties of arc sprayed fibers and composites. NASA CP-10104, 31.1– 31.17.

Saravanos, D. A. (1993). Mechanics for the effects of delaminations on the dynamic characteristics of composite laminates. In *Dynamic Characteristics of Advanced Materials* (Edited by P. K. Raju and R. F. Gibson), ASME Winter Annual Meeting, NCA-16/AMD-172, pp. 11–21. ASME, New York.

Sayir, H., Sayir, A. and Lagerlzöf, K. P. D. (1993). Temperature-dependent brittle fracture of undoped and impurity-doped sapphire fibers. *Ceramic Engng Sci. Proc.* **14**, 581–590.

Serafini, T. T., Delvigs, P. and Lightsey, G. R. (1972). Thermally stable polyimides from solutions of monomeric reactants. *J. Appl. Polymer Sci.* **16**, 905.

Serafini, T. T., Vannucci, R. D. and Alston, W. B. (1976). Second generation PMR polyimides. NASA TMX-71984.

Serrano, A. M., Jangchud, I., Eby, R. K., Bowles, K. J and Jayne, D. T.(1994). Interfacial bonding mechanisms in carbon fiber/PMR-15 composites. In *MRS Symp. Proc.* no. 305, *High Performance Polymers and Polymer Matrix Composites* (Edited by R. K. Eby, R. C. Evers, M. A. Meador and D. Wilson), pp. 105–111. Materials Research Society.

Singh, M., Dickerson, R. M. and Behrendt, D. R. (1994). Characterization of silicon carbide fiber reinforced reaction-formed silicon carbide matrix composites. *18th Annual Conf. on Advanced Ceramics and Composites*, Cocoa Beach, FL. To be submitted for publication in *Ceramic Engng Sci. Proc.*

Stephens, J. R. (1990). NASA's HITEMP program for UHBR engines. AIAA-902395.

Stephens, J. R., Hecht, R. J. and Johnson, A. M. (1993). Material requirements for the high speed civil transport. *AIAA Conf. Proc., International Symp. on Air Breathing Engines*, Japan. AIAA, Washington, DC.

Thomas, D. J. and Wetherhold, R. C. (1991). Reliability analysis of laminates with load sharing. *J. Compos. Mater.* **25**, 1459–1475.

Tiwari, A. and Henneke, E. G., II (1993). Real time acousto-ultrasonic NDE technique for monitoring damage in SiC/CAS ceramic composites. *Proc. 2nd International Conf. on Acousto-ultrasonics: Acousto-ultrasonic Materials Characterization* (Edited by A. Vary), pp. 273–283. American Society for Nondestructive Testing.

U. S. Department of Health and Human Services (1986). Methylenedianiline. *Current Intelligence Bulletin* 47.

Vannucci, R. D. (1987). PMR polymide compositions for improved performace at 371°C. *SAMPE Q.* **19**, 31.

Vannucci, R. D. and Chriszt, J. K. (1993). Properties of MDA-free polyimide composites. NASA CP-19117, 12-1–12-11.

Vannucci, R. D., Malarik, D. C. and Waters, J. F. (1990). Autoclavable addition polyimides for 371°C composite applications. *Proc. 22nd International SAMPE Technical Conf.*, Boston, MA. SAMPE, Covina, CA.

Vannucci, R. D., Olshavsky, M. A. and Chriszt, J. K. (1992). Processability and properties of blocked PMR polyimide resins and composites. NASA CP-10104, 9-1–9-15.

Vary, A. (1990). Acousto-ultrasonics. In *Non-destructive Testing of Fiber Reinforced Plastics Composites* (Edited by J. Summerscales), Vol. 2, pp. 18–54. Elsevier Applied Science, Barking.

Vary, A. (1992). NDE standards for high temperature materials. In *Nondestructive Testing Standards—Present and Future*, ASTM STP 1151 (Edited by H. Berger and L. Mordfin), pp. 211–224. ASTM, Philadelphia, PA.

Waters, J. F., Sutter, J. K., Meador, M. A. B., Baldwin, L. J. and Meador, M. A. (1991). Addition curing thermosets endcapped with 4-Amino[2.2]paracyclophane. *J. Polymer Sci.*: Part A: *Polymer Chem.* **29**, 1917–1924.

Ye, P. (1994). The failure statistics of an oxide-oxide ceramic composite. M. S. Thesis, Cleveland State University.

Yun, H. M. and Goldsby, J. C. (1993). Tensile creep behavior of polycrystalline alumina fibers. NASA TM 106269.