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.

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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.

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Another development of equal significance was the adaptation of powder metallurgy processing techniques, such as hot isostatic pressing and superplastic forging, to the production of turbine disks. Along with these developments, which were focused primarily on the superalloys, important gains in processing capabilities were also achieved for titanium alloys. At the same time, advanced composites were developed, and these have found useful applications in the engine.

The situation today is that superalloys account for about 50 percent by weight of advanced engines, with the balance roughly equally divided between titanium alloys, composites, and steels (Fig. 1). The trend appears to be in the direction of somewhat higher weight fractions of superalloys and composites, at the expense of titanium alloys and steels. In the superalloy area, the trend is toward increasing applications for directionally solidified (DS) and powder metallurgy (PM) products.

Summary. Materials and processing innovations that have been incorporated into the manufacture of critical components for high-performance aircraft gas turbine engines are described. The materials of interest are the nickel- and cobalt-base superalloys for turbine and burner sections of the engine, and titanium alloys and composites for compressor and fan sections of the engine. Advanced processing methods considered include directional solidification, hot isostatic pressing, superplastic forging, directional recrystallization, and diffusion brazing. Future trends in gas turbine technology are discussed in terms of materials availability, substitution, and further advances in air-cooled hardware.

This article presents an overview of materials developments as they are related to specific components in the gas turbine engine, such as blades, vanes, disks, and combustors. A brief description will first be given of the design and operation of component parts in an engine to provide the reader with a better appreciation of the factors involved in materials selection.

Engine Components

Gas turbine engines such as the JT9D turbofan engine (Fig. 2) comprise three main sections: compressor (fan), burner, and turbine (*1*). The compressor raises the pressure and temperature of the incoming air and delivers it to the burner. In the combustion chamber, a fine spray of fuel is thoroughly mixed with the high-pressure air, and the mixture is ignited. The hot gases leaving the combustion chamber undergo rapid expansion in the turbine section, accompanied by a sharp drop in gas pressure and temperature. The effect is to drive the turbine, which in turn drives the compressor. The high-velocity gases expelled through the exhaust nozzle generate engine thrust. Additional thrust is derived from the relatively low-velocity bypass air driven by the fan and ducted outside the engine.

The indicated temperature and pressure variations correspond to maximum thrust developed on takeoff and represent important design parameters. From the materials standpoint, the crucial factor is the peak temperature developed at different locations in the engine, since this sets an upper limit on the required temperature capability of the material. Other materials design considerations of equal significance are (i) the magnitude of stresses developed by centrifugal forces, vibratory forces, and thermal gradients and (ii) the potential for oxidation (or hot corrosion) and erosion in the high-velocity gas stream.

Fan. The fan section of the engine is integral with the front part of the low-pressure compressor. This permits the fan blades to rotate at low tip speed, consistent with optimum fan efficiency. The fan blades are relatively long and thin components and are braced (shrouded) at the midspan for support and to prevent vibration. Air passing through the fan and exhausted through the ducts tends to carry away ingested foreign material, which otherwise might damage the engine core.

Materials for fan blades must be strong, elastically stiff, and resistant to damage by foreign objects. Experience has shown that titanium alloys satisfy these requirements. Composites with high specific strength, such as carbon-epoxy, have also been considered for this application, but at present lack sufficient resistance to foreign object damage.

Compressor. The compressor consists of a series of rotating blades and stationary vanes, which are arranged in stages concentric with the axis of rotation. The blades are fastened into slots around the periphery of individual disks; the vanes are attached to the compressor casing between adjacent rotor stages. The flow path, defined by successive stages of the compressor, decreases in the direction of air flow, corresponding to the reduction in volume as compression occurs from stage to stage (Fig. 2). The high-pressure compressor runs hotter and at higher speeds than the low-pressure compressor.

Blades and vanes must be capable of resisting aerodynamic loading. In addition, blades must be able to withstand high centrifugal loading and the effect of vibratory stresses caused by high-velocity air streaming from the spaces between blades. Disks must possess high load-bearing capacity, since they have to resist the centrifugal loading of the disk and blades. Some consideration must also be given to the effects of high strains developed at the critical points of attachment of the blades to the disk. Steels and titanium alloys have proved to be satisfactory materials for blades, vanes, and disks in the low-pressure compressor, but the more heat-resistant nickel-base alloys are preferred for the hotter sections of the high-pressure compressor.

Burner. The burner section is essentially an annular combustion chamber, in which several burner cans are arranged side by side around the inside of the chamber. Each burner can, which is a separate combustion chamber, is perforated with holes and slots to admit air for cooling. About one-third of the total volume of air entering the burner from the compressor discharge is mixed with the fuel to sustain combustion. The remainder of the air bypasses the fuel nozzles and is used downstream to cool the burner can surfaces and combustion products before the hot gases enter the turbine. The system is designed to maintain an even temperature distribution in the hot gases leaving the combustor and to ensure that the peak temperature does not exceed the allowable limit at the turbine inlet.

Burner materials must be formable, weldable, and resistant to corrosion, distortion, and thermal fatigue at high temperatures. The strength requirements are relatively low, but the strength must be maintained up to approximately 1100°C. A special class of heat-resistant sheet alloys have been developed for this purpose.

Turbine. The configuration of the main components of the turbine—the rotor and interstage guide vanes—is similar to that of the compressor, except that the gas path increases in the direction of flow to accommodate the expansion of gases between stages in the turbine. The blades are attached to the disks, using a characteristic fir-tree root design, which leaves space for expansion. Vanes are slotted into the turbine casing in stages between the rotors. To reduce vibrations and to increase turbine efficiency, the blade tips are sometimes shrouded. To permit a higher turbine inlet temperature, which is critical to improve operating efficiency, inlet guide vanes and first-stage blades are air-cooled. This is accomplished by passing compressor bleed air through longitudinal holes, tubes, or cavities in the airfoil sections (Fig. 3). The cooling air exits through small holes and slots at the leading and trailing edges of the airfoil. Air cooling is necessary only in the turbine inlet section, since the energy extracted from the hot gases by the first- and second-stage rotors reduces the temperature to a tolerable level. It is illustrative of the effectiveness of air cooling that even with a turbine inlet temperature as high as 1300°C, metal temperatures for inlet guide vanes and first-stage blades are easily maintained at 1100° and 950°C, respectively.

The materials requirements for turbine blades and vanes are similar, except for the reduced strength requirements in vanes. Stresses in vanes are less than 35 megapascals, whereas blades are subjected to longitudinal stresses up to 200 MPa in the midspan of the airfoil section. The blade root, which is attached to the disk, is outside the gas path and experiences a maximum temperature of approximately 750°C, but tensile stresses are much higher, in the range 275 to 550 MPa. The primary requirements are creep strength at high temperatures, thermal fatigue resistance, and resistance to environmental degradation by oxidation, corrosion, and erosion. Secondary requirements include castability, impact strength, and microstructural stability to ensure that properties are maintained for long periods of time. Some vane designs also require weldability. Turbine disks operate at temperatures that do not exceed approximately 750°C. The maximum temperature occurs at the outer edge or rim of the disk. The stresses developed by centrifugal loading are high at the rim and even higher in the bore. Primary materials requirements are for high burst strength at the operating temperature of the bore and good creep strength at rim operating conditions. The materials should also possess good fatigue strength, both low-cycle and high-cycle. A bewildering variety of superalloys, both nickel- and cobalt-base, have been designed for turbine applications.

Materials and Applications

Superalloys. The high-temperature alloys (*2–5*), known as the superalloys, occur in two broad classes: nickel-base and cobalt-base (Table 1). Nickel-base alloys are strengthened by precipitation of the $Ni_3(Al,X)$ γ' phase, where X is a solid solution hardening element, such as Ti, Nb, or Ta. In most advanced alloys, γ' precipitation hardening is supplemented by solid solution hardening of the γ matrix with refractory elements, such as W, Mo, or Re. The γ' particles precipitate on a fine scale (Fig. 4a), and both precipitate and matrix phase are coherent. This is a consequence of the close similarity in structure between the face-centered cubic γ and ordered face-centered cubic γ' phases. The cobalt-base alloys derive their strength from combined solid solution hardening and carbide dispersion strengthening. Fine networks of carbide phases (Fig. 4b) appear to be particularly effective in strengthening at high temperatures.

The nickel-base alloys have outstanding strength characteristics at temperatures in the range of 750° to 1000°C. The cobalt-base alloys are stronger and more corrosion-resistant at temperatures above about 1050°C. Accordingly, the nickel-base alloys are used for intermediate-temperature blade and disk ap-

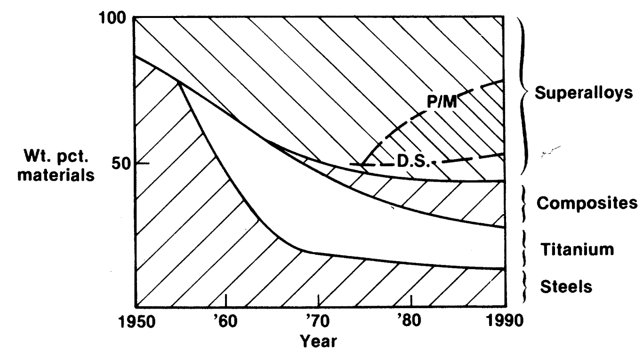

Fig. 1. Weight percent of materials used in most advanced aircraft gas turbine engines.

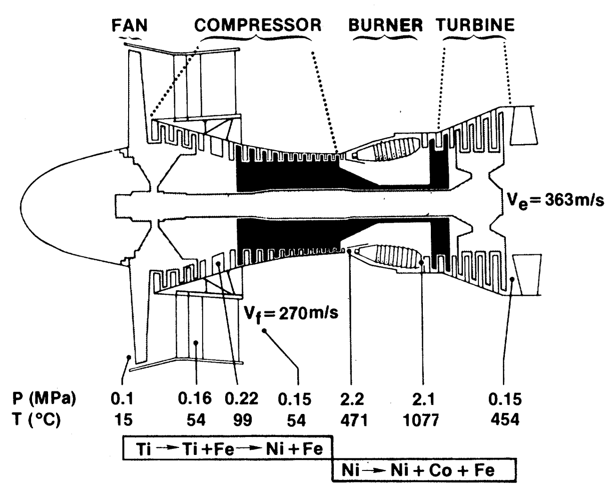

Fig. 2. Internal pressure (*P*) and temperature (*T*) variations in the Pratt & Whitney JT9D Turbofan engine at a sea-level takeoff thrust of 19,500 kilograms and with a bypass ratio of 5:1.

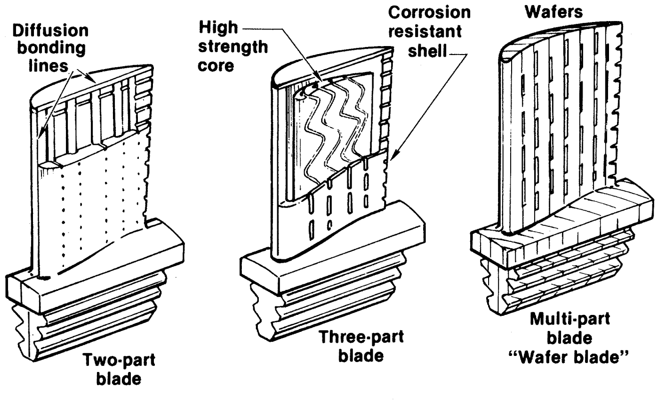

Fig. 3. Advanced air-cooled turbine blades.

Table 1. Nominal compositions of some nickel-base and cobalt-base alloys.

Alloy	Composition (% by weight)								
	Ni	Cr	Co	C	Ti	Al	Mo	W	Other
				Blade alloys					
Waspaloy	Bal.*	19.5	13.5	0.08	3.0	1.4	4.0		0.08 Zr, 0.007 B
B-1900	Bal.	8.0	10	0.11	1.0	6.0	6.0		0.07 Zr, 0.015 B, 4.3 Ta
PWA 1422	Bal.	9.0	10	0.11	2.0	5.0		12.5	0.015 B, 1.0 Cb, 2.0 Hf
				Disk alloys					
Incoloy 901	Bal.	12.5		0.10	2.6		6.0		34 Fe, 0.015 B
Waspaloy	Bal.	19.5	13.5	0.08	3.0	1.4	4.0		0.08 Zr, 0.007 B
IN-100	Bal.	12.4	18.5	0.07	4.3	5.0	3.2		0.06 Zr, 0.02 B, 0.8 V
				Vane alloys					
X-40	10	25	Bal.	0.50				7.5	1.5 Fe
WI-52		21	Bal.	0.45				11	1.75 Fe, 2.0 Cb
MAR-M509	10	24	Bal.	0.60	0.2			7	0.5 Zr, 3.5 Ta
				Burner alloys					
Hastelloy X	Bal.	22	1.5	0.10			9	0.6	18.5 Fe
Haynes 188	22	22	Bal.	0.10				14.5	0.08 La
Inconel 617	Bal.	22	12.5	0.07		1.0	9		

*Balance.

plications, whereas the cobalt-base alloys are generally preferred for high-temperature vane applications.

The essential requirement for blade alloys is adequate creep strength at elevated temperatures. Experience has shown that this requirement is best satisfied by exploiting the nickel-base alloys with a high γ' volume fraction. In the 1950's, alloys of this type, containing about 30 percent γ', were fabricated into blades by hot forging (Fig. 5). In the 1960's following the development of improved alloys containing up to 60 percent γ', forging was replaced by investment casting as the preferred method of blade fabrication. Later this was to evolve into directional solidification processing of blades, including both columnar-grained and single-crystal materials. Looking at the overall picture, it can be seen that such processing innovations, along with advances in alloy design, have been responsible for an increase of about 150°C in the permissible operating temperature of blades. Current interest in this area is focused on exploiting the higher strength capabilities of directionally solidified eutectics (*6*) and directionally recrystallized (*7*) conventional superalloys. The eutectic microstructure is composite in nature, consisting of a γ/γ' matrix and reinforcing fibers, or lamellae of a refractory phase. An example is the γ/γ'-α eutectic, which is reinforced with thin filaments of α-Mo (Fig. 4c). Such alloys promise to increase the temperature capability of blade alloys by a further 75°C in the 1980's (Fig. 5).

At one time, a clear distinction could be made between blade and disk alloys. Blade alloys were selected for intermediate-temperature creep strength, disk alloys for high strength at somewhat lower temperatures. As operating temperatures in the engine have increased, this distinction has become blurred because of the need to provide additional creep strength in the disk alloys. The solution to this problem has been to increase the γ' volume fraction of disk alloys at the expense of hot workability. As a consequence, conventional hot forging of disks, which is limited to workable low γ' alloys, has been gradually replaced by more flexible PM processing methods (Fig. 6). The new processes, which are known as superplastic forging and hot isostatic pressing, are applicable to all nickel-base alloys. Although appreciable gains in strength have been achieved with the new alloys and processing methods, this benefit is perhaps not as significant as the improvement in creep strength, because of the higher rim temperatures encountered in today's engines. For this reason, another likely development in the 1980's is the dual-property disk, in which the bore of the disk is optimized for load-bearing capacity and the rim for creep strength.

Cobalt-base alloys have found their widest application as vane materials. They are attractive for this application because they can be processed by relatively inexpensive investment casting techniques, without having to resort to vacuum melting. The early alloys, such as X-40 and WI-52 (Fig. 7), are examples of alloys in this air-melting category. For newer alloys, such as MAR-M509, which

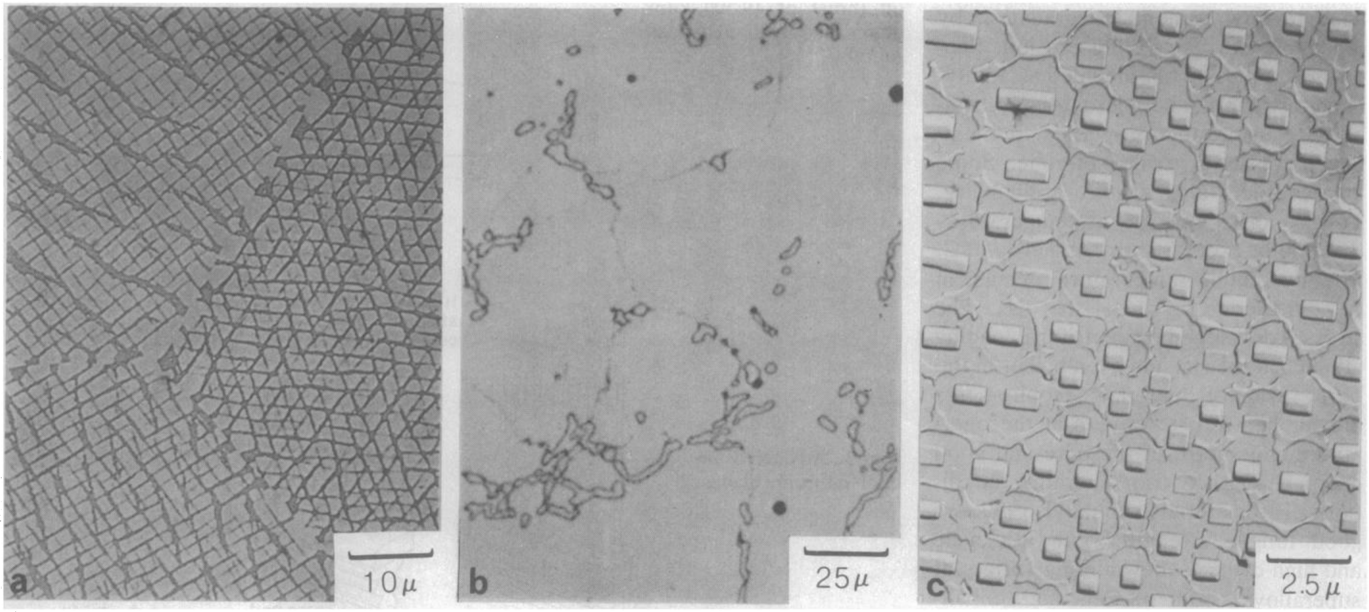

Fig. 4. Representative microstructures of (a) γ'-strengthened nickel-base alloy, (b) carbide-strengthened cobalt-base alloy, and (c) fiber-strengthened γ/γ'-α (Mo) eutectic alloy.

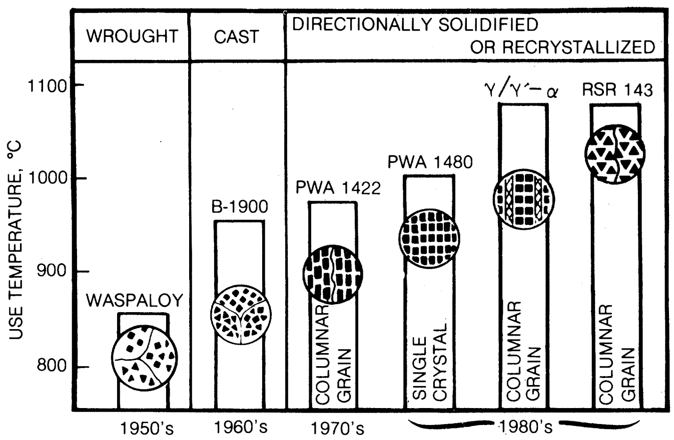

Fig. 5 (left). Evolution of blade alloys.

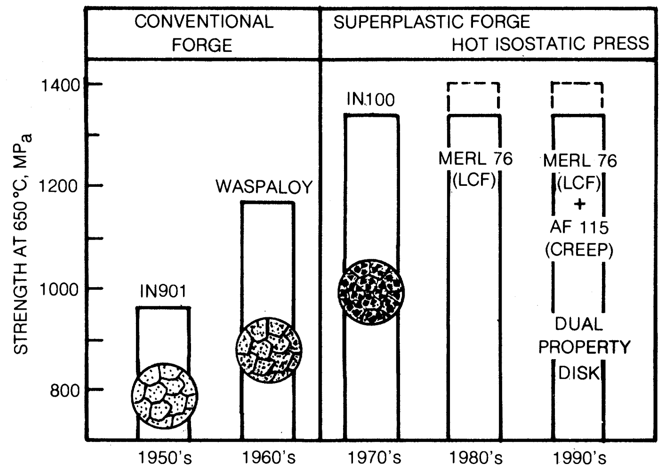

Fig. 6 (right). Evolution of disk alloys.

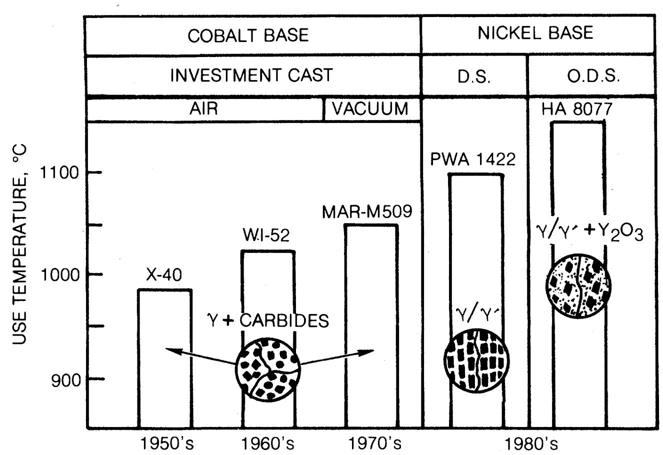

Fig. 7. Evolution of vane alloys.

contain more reactive elements, vacuum melting and casting is necessary. At present, the future of cobalt-base alloys in engines is uncertain, because of current and projected shortages of cobalt. It seems clear, however, that greater use will be made of directional solidification of vane alloys, because of the real improvements in thermal fatigue resistance exhibited by DS structures. It also seems likely that some use will be made of the new generation of oxide dispersion strengthened (ODS) alloys, which are now becoming available commercially. A lot will depend on the cost of fabricating these materials into vane shapes. These new alloys combine high-temperature strength with remarkable resistance to oxidation and hot corrosion.

Combustor components are fabricated from sheet alloys by a variety of conventional shaping and joining operations. High workability and good weldability are essential for burner alloys, which precludes the use of precipitation-hardenable systems. Experience has shown that the necessary strength at high temperatures can be achieved by refractory metal (W or Mo) solid solution strengthening of either nickel- or cobalt-base alloys. High Cr levels are necessary to maintain adequate resistance to oxidation and hot corrosion. Representative alloy compositions for service up to ~ 1100°C are given in Table 1.

Distortion of combustors due to thermal cycling and local hot spots remains a problem. Attempts are being made to improve the situation by exploiting the more heat-resistant ODS alloys, such as HA 8077 (NiCrAl with Y_2O_3 dispersion). The ODS alloys have the potential for use at an ~ 100°C higher temperature than conventional sheet alloys. Manufacture of ODS sheet products has been demonstrated, but costs are high. Improved manufacturing techniques, such as mechanical alloying, promise substantial reduction in the cost of these materials (*8, 9*).

Titanium alloys. Titanium alloys are categorized by the microstructural phases that predominate near room temperature (*10, 11*). Titanium exhibits an allotropic phase transformation; β, the high-temperature phase, is body-centered cubic, and α, the low-temperature phase, is close-packed hexagonal. In pure titanium this transformation occurs at 882°C. Alloying elements may act to stabilize either the β or the α phase. Aluminum is the most important of the α-stabilizing elements. The β alloys have low creep strengths at elevated temperatures and are therefore limited to low-temperature applications. Table 2 lists selected properties of three conventional titanium alloys that are representative of current compressor disk and blade materials.

Materials used for moderate-temperature applications are often the heat-treatable $\alpha + \beta$ alloys, such as Ti-6Al-4V, with fine-grained microstructures of α and transformed β. The alloys selected for high-temperature compressor and disk components may be near-α or α-lean β alloys. Examples of these respective alloys are Ti-8Al-1Mo-1V, which contains small amounts of the β-stabilizing elements Mo and V, and Ti-6Al-2Sn-4Mo-2Zr, where solid solution strengthening of the α phase by Al, Sn, and Zr produces improved creep resistance. The relatively good strength of Ti-8Al-1Mo-1V and its higher elastic modulus and lower density, as indicated in Table 2, favor its use for applications where material stiffness is important.

The application of titanium alloys to the blades and disks of the compressor section of the engine has been a major contributor to its evolution. This usage was prompted by the low density of titanium alloys, which leads to superior specific strength of these alloys at temperatures up to about 500°C. In rotating gas turbine components, tensile, creep, low- and high-cycle fatigue, fracture toughness, and erosion properties are the important selection criteria. Alloying and thermomechanical processing are used to achieve an appropriate balance of these properties and to improve the temperature capability of titanium alloys.

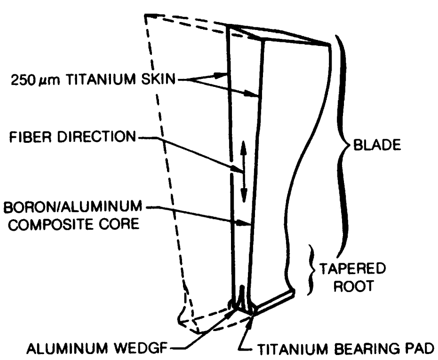

Fig. 8. Composite fan blade.

Despite the improvements that have been made, the upper use temperature for titanium alloys is disappointingly low relative to the melting point of titanium. Some promise for overcoming this temperature limitation is provided by the Ti_xAl_y-based intermetallics, TiAl and Ti_3Al. These intermetallics combine low densities with a high-temperature capability. They show loss of ductility above room temperature, and this problem is being addressed. Another approach, which may further enhance the specific properties of titanium alloys and their upper use temperature, involves the use of fibers, such as silicon carbide, for reinforcement. The eventual application of such composite systems awaits extensive materials development, component design, and evaluation efforts.

The initial developments in titanium alloys were not without significant problems. Some of these, such as alloy embrittlement in the presence of hydrogen concentrations higher than 150 parts per million and alloy element segregation and contamination, were solved by careful vacuum arc melting. Titanium alloys are now commonly produced by multiple consumable electrode arc melting under vacuum. This procedure degasses and homogenizes the cast structure.

Gas turbine components of titanium alloys are usually produced from hot-forged and heat-treated products. The properties of these alloys depend on microstructural morphology as well as composition. The morphology is subject to the thermomechanical processing history, and microstructural variations that occur in some products, in turn, result in variability of properties. This sensitivity is a consequence of effects such as that of cooling rate on the β-to-α transformation. Advances have been made in the fabrication of titanium alloys by PM techniques, and this approach promises to improve their compositional and microstructural homogeneity.

Composites. The high strengths, high stiffnesses, and low densities characteristic of high-performance fiber-reinforced materials make them attractive candidates for gas turbine structural applications (*12*) (Table 3). The most important current composite systems are graphite/epoxy, graphite/polyimide, and boron/aluminum. Graphite/epoxy is finding application as a material for fan exit guide vanes. This system has a use temperature up to 175°C. The graphite/polyimide system, with a use temperature to 325°C, is being evaluated for exhaust flaps in an advanced military engine. Another component subject to substantial development activity is the fan blade of boron/aluminum (Fig. 8). In comparison to the present titanium blade, the composite blade is capable of a higher tip speed and may, because of its higher modulus, be designed without the midspan stiffener, which leads to improved aerodynamic efficiency. The 1980's should see increased applications of composites as engine components. Cost and reliability will be important considerations for these applications.

Table 3. Properties of reinforcing fibers.

Fiber	Strength (MPa)	Elastic modulus (GPa)	Density (kg/m^3)
Graphite	2100-2400	200-400	1700-1900
Boron	3500	400	2500
Silicon carbide	3450	400	3000

Table 2. Selected properties of representative titanium alloys.

Alloy	Density (kg/m^3)	Annealing/solution treatment temperatures (°C)	20°C elastic modulus (GPa)	Typical strength (MPa)	
				20°C	538°C
Ti-6Al-4V	4428	732-760/899-968	11.3	1000	480
Ti-8Al-1Mo-1V	4373	760-788/non-heat-treatable	12.7	1000	630
Ti-6Al-2Sn-4Mo-2Zr	4539	704-843/829-913	11.3	1000	680

Coatings. The development of coatings to extend the life of a gas turbine airfoil began in about the mid-1950's. However, it was not until the late 1950's that the first practical application to cobalt-base alloy vanes was realized. Since then, successful applications of coatings have been made to blades, burner cans, and other critical components in the gas turbine engine (*13*).

The earliest coating methods involved some type of aluminizing treatment; that is, diffusion of aluminum into the surface layers of the alloy substrate. The effect of such a treatment on nickel- and cobalt-base alloys is to produce protective layers consisting mainly of the intermetallic compounds NiAl and CoAl (Fig. 9a). These compounds impart good oxidation and hot corrosion (sulfidation) resistance, because they form a continuous impervious scale of alumina on exposure to a high-temperature oxidizing environment. Two coating methods have been widely used for aluminizing superalloys: slurry fusion and pack cementation. In the slurry process, the part is sprayed with a suspension of aluminum or aluminum alloy and subjected to a high-temperature treatment to produce melting and interdiffusion between deposit and substrate. In the pack cementation process, the part is reacted with aluminum or aluminum alloy powder in the presence of an ammonium halide activator at an elevated temperature. The operation is carried out in an hermetically sealed container to maintain a controlled activity of the aluminum in the vapor phase. Typical coating thicknesses are in the

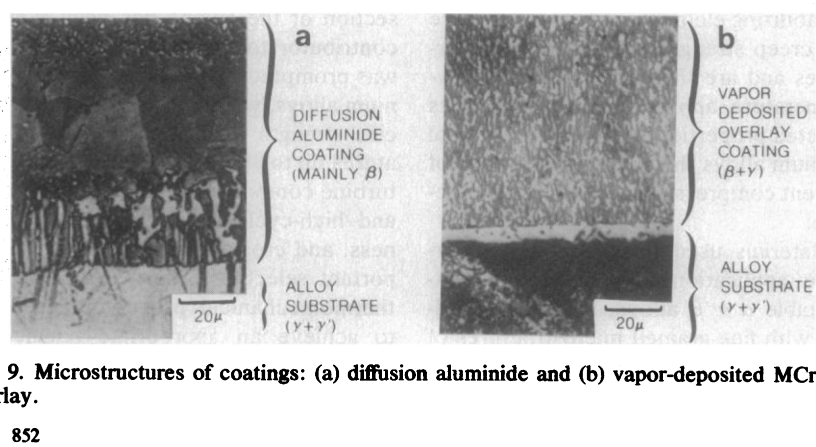

Fig. 9. Microstructures of coatings: (a) diffusion aluminide and (b) vapor-deposited MCrAlY overlay.

range 25 to 100 micrometers. Advanced coatings treatments include additions of Cr and Pt to further enhance the oxidation and hot corrosion resistance of the coating. Major improvements in hot corrosion resistance can be achieved by proper control of the Pt distribution in the aluminide layer. The Pt-rich coatings are synthesized by electroplating a thin (about 5 μm) layer of Pt before aluminizing.

A limitation of the diffusion aluminide coatings is that they seriously lack ductility at temperatures below about 750°C. Thus, they are highly susceptible to surface cracking under thermal cycling conditions. Another problem is the relatively poor adherence of the protective oxide scale to the coating alloy. Under thermal cycling conditions, it is not uncommon to encounter repetitive detachment or spallation of the alumina scale, leading to rapid degradation of the coating due to loss of Al. To resolve these problems, the emphasis in coating technology shifted in the mid-1960's to direct bonding or cladding of optimized coating compositions to the superalloy substrate. Ductility was improved by making adjustments in coating compositions so as to develop microstructures in which the brittle β-NiAl or β-CoAl is embedded in a ductile γ solid solution matrix. Oxide adherence was improved by making trace additions of rare earth elements, such as yttrium. The final result of this work was the introduction of the MCrAlY coatings, where M is Ni, Co, or Ni + Co (Fig. 9b). As in the case of diffusion coating, overlay coatings also benefit from additions of Pt.

Many methods of processing MCrAlY overlay coatings have been tested, including diffusion bonding, powder sintering, plasma spraying, vacuum evaporation, sputtering, and ion plating. Of these, electron beam evaporation has emerged as the preferred method of coating superalloy blades and vanes. In a typical arrangement, using continuous ingot feed, it is necessary to impart a complex motion to the part in order to achieve a uniform coating deposit.

An oxide scale on the alloy coating acts to some extent as a thermal barrier, because of its relatively poor heat transfer characteristics. Its effectiveness, however, is limited because its thickness is typically < 1 μm. To improve the situation, the concept of applying thermal barrier coatings, on top of existing coatings, has emerged (*14*). So far the method has been applied successfully to sheet metal components, such as burner cans and exhaust liners. The coatings are based on yttria-stabilized zirconia, and are applied by plasma-spraying layers 125 to 500 μm thick. To minimize the effects of stresses due to thermal expansion mismatch between alloy substrate and zirconia deposit, coatings compositions are normally graded—that is, the composition gradually changes from base metal to pure zirconia over the coating thickness. Thermal barrier coatings are now being developed for turbine airfoils, to reduce metal temperatures and thereby reduce cooling-air requirements. Some of the best results have been obtained with a 250-μm plasma-sprayed coating of yttria-stabilized zirconia over a 100-μm layer of NiCrAl coating (Fig. 10). Metal temperature reductions of ~ 100°C have been achieved in an engine test of experimental air-cooled vanes. Increased applications of ceramic coatings for thermal insulation and as abradable seals are expected in the 1980's.

Advanced Processes

Directional solidification. In 1960 it was demonstrated that the creep properties of superalloys can be improved markedly by directional solidification to align all the grain boundaries parallel to the direction of the applied tensile stress (*15*). The effect was attributed to the elimination of grain boundaries transverse to the applied stress, since such boundaries are highly susceptible to cavitation and cracking under creep conditions. In 1967 it was shown that further improvements in properties could be achieved by eliminating the grain boundaries altogether; that is, by directional solidification to produce single crystals (*16*). Following these leads, in the late 1960's, Pratt & Whitney Aircraft developed various commercial processes for the directional solidification of bars, ingots, slabs, and more complex shapes. The most notable achievement was the directional solidification of cast-to-size turbine blades (*17*) (Fig. 11).

The DS process evolved from conventional investment casting practice. In conventional casting, the melt is poured into a ceramic shell mold, and solidification is allowed to occur in a relatively uncontrolled manner by radiation to the walls of the vacuum chamber (Fig. 12a). Since the mold preheat is usually about half the melt temperature, the melt experiences a chilling effect in the mold, which results in fairly rapid solidification and the formation of a fine polycrystalline structure. In directional solidification the situation is similar, except for a higher mold preheat temperature and some provision for controlling temperature gradients in the mold. This is accomplished by attaching the ceramic mold, open at its base, to a water-cooled copper chill plate. Following the introduction of the melt into the mold and the commencement of solidification in the chill zone, the temperature gradient

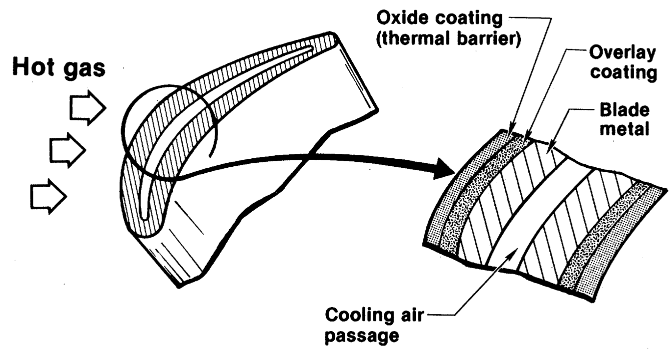

Fig. 10. Thermal barrier coating.

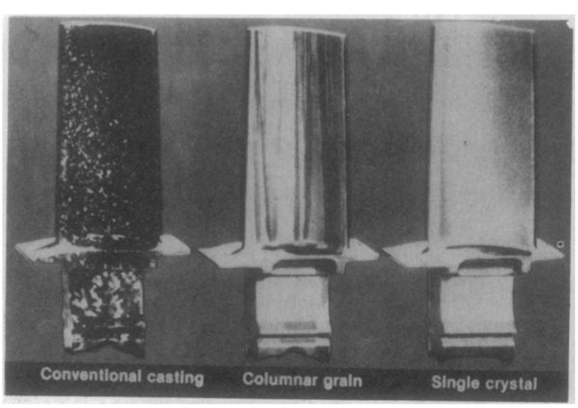

Fig. 11. Effect of processing on grain structure of cast-to-size turbine blades.

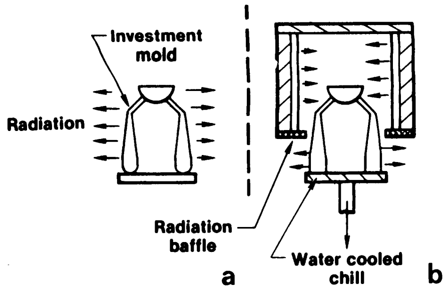

Fig. 12. Comparison of (a) conventional casting and (b) directional solidification.

in the mold is adjusted to promote the advance of the solid-liquid interface in a direction normal to the chill surface. Thus, a characteristic directionally aligned or columnar-grained structure is developed in the casting. A useful consequence of the steep temperature gradients developed in the chill zone is the formation of a strong <100> or cube texture, which is incorporated into the columnar grain structure. Current practice favors the arrangement depicted in Fig. 12b. The desired temperature gradient is maintained by gradually lowering the mold from the hot zone, through an array of heat shields, into the vacuum chamber, where radiation losses can occur freely.

The procedure for processing single-crystal superalloys emerged quite naturally from the basic DS process. A crystal selector is incorporated into the mold just above the chill zone. After developing the desired <100> texture in the chill zone, the crystal selector is used to exclude all but one grain, which then expands to fill the mold. To obtain single crystals in any desired orientation, conventional seeding techniques must be used in conjunction with the DS process. The oriented seed crystal is attached to the base of the mold in contact with the chill plate. The maximum allowable growth rate for the DS processing of single crystals is 500 centimeters per hour, while columnar-grained material can be produced at rates up to 40 cm/hour.

Nickel-base superalloys respond well to DS processing. This applies to both conventional γ' precipitation-hardened alloys and eutectic superalloys, such as γ/γ'-α (Mo) and γ/γ'-TaC. Eutectic superalloys, however, require steeper temperature gradients and much lower growth rates (typically ~ 2 cm/hour) to obtain optimally aligned structures. Significant improvements in creep rupture properties by DS processing have been obtained for both types of superalloys (Fig. 5). Conventional superalloys also exhibit improved thermal fatigue properties in the <100> orientation. These property advantages have been exploited in DS superalloy castings of first- and second-stage turbine blades and inlet guide vanes. At present, superalloy castings of the columnar grain type are the most widely used in commercial engines. However, the situation is changing, and it appears that before long single-crystal castings will displace them for the most critical applications in the engine. This is because of the higher temperature capabilities that have been achieved in a new class of superalloys, designed specifically to exploit the single-crystal character of the material. An example of an alloy in this category is PWA 1480, which offers a 25°C advantage over its columnar-grained predecessor, PWA 1422 (Fig. 5). It remains to be seen whether the demonstrated advantages in creep strength of certain eutectic superalloys, such as γ/γ'-α, will offset the higher manufacturing costs due to the lower growth rates.

In contrast to these developments, the situation with respect to cobalt-base superalloys for vane applications has remained relatively static. As indicated in Fig. 7, conventional investment casting has maintained a dominant position in the processing of cobalt-base superalloys since the 1950's, although the temperature capabilities of the alloys have continued to improve. The picture may be changing because of continuing shortages in the supply of cobalt. It now seems likely that the 1980's will see the gradual adoption of DS nickel-based superalloys, such as PWA 1422, for vane applications, along with ODS nickel-base alloys, such as HA 8077.

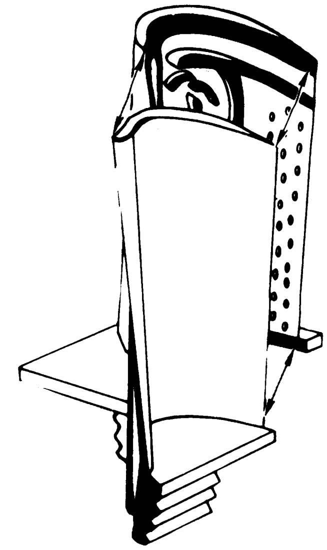

Fig. 14. Diffusion brazing of matched blade halves.

Superplastic forging. In 1963 it was found that fine-grained superalloys exhibit high ductility or superplastic behavior when deformed at high temperatures under appropriately low strain rate conditions (*18*). This paved the way for the subsequent development of a new process for the hot deformation processing of superalloys, which has become known as the Gatorizing forging process.

The desired fine-grained microstructure can be obtained most conveniently by hot extrusion. Typically, the material is subjected to high deformation rates at temperatures just below the γ' solution temperature. This is a crucial aspect of the process, since the high deformation rate causes spontaneous recrystallization on a very fine scale and the presence of the γ' particles serves to stabilize the fine grain structure. The resulting grain size is in the range 1 to 5 μm, depending on the extrusion ratio and temperature. When the optimum fine-grained structure has been achieved, the material may be worked into shape by exploiting its superplastic character. This is accomplished by controlled strain rate, isothermal forging—that is, superplastic forging. After forging, the creep strength of the material is recovered by a solution heat treatment to coarsen the grain structure, followed by quenching and aging to obtain the opti-

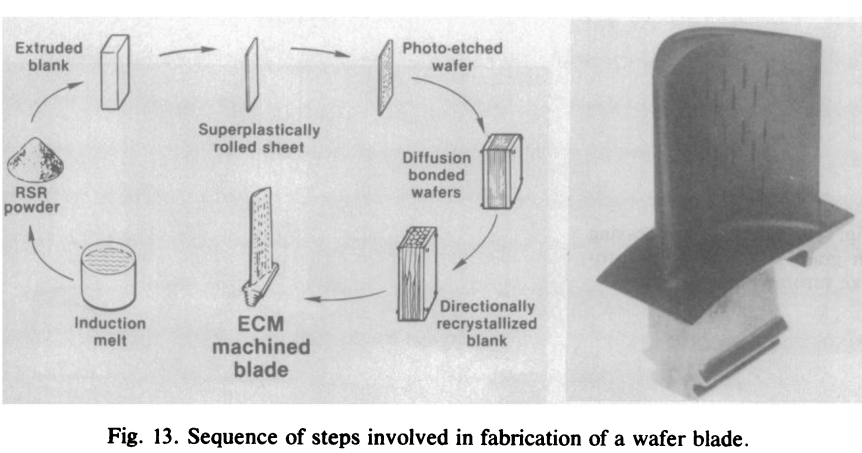

Fig. 13. Sequence of steps involved in fabrication of a wafer blade.

mum distribution of the γ' hardening phase.

Superplastic forging has been used to fabricate superalloy turbine disks and even complete rotors starting from prealloyed powder. A particular advantage of the process, compared with conventional hot forging, is its ability to generate a forged profile that more closely matches the final shape of the article. This reduces machining costs, because less material has to be removed in the finish machining operation.

A characteristic feature of the fine-grained superplastic material, at least when produced from prealloyed powder, is its remarkable chemical and microstructural uniformity. Another favorable aspect is its ability to form a directionally aligned or columnar-grained structure when subjected to directional recrystallization (DR) under the influence of a steep thermal gradient (*19*). The resulting grain structure bears a superficial resemblance to that obtained by directional solidification. However, there are important distinctions. The DR grain structure is finer in scale and much more homogeneous than the DS structure. Furthermore, the DR structure exhibits not one, but several grain textures, including, <100>, <110>, and <111>, depending on processing parameters and alloy composition.

This ability to develop textured columnar-grained structures entirely by solid-state processing has stimulated interest in new methods for the fabrication of air-cooled blades and vanes of advanced design, starting from superplastically formed sheet stock (*20, 21*). One such processing scheme is depicted in Fig. 13. The critical step in the process is the diffusion bonding of thin photoetched wafers to produce the desired configuration of internal cooling passages. Following directional recrystallization of the bonded structure, the actual blade profile is formed by electrochemical machining. An experimental air-cooled wafer blade, similar to that shown in Fig. 13, has already withstood higher turbine inlet temperatures than even the most sophisticated one-piece or two-piece blade castings (Fig. 3). This is because of the much more efficient air-cooling schemes that are attainable with the multiple-wafer fabrication technique.

Hot isostatic pressing. In 1955, a process was invented for the gas-pressure diffusion bonding of materials, which later became known as hot isostatic pressing (HIP) (*22, 23*). After 20 years, the process has found its greatest commercial significance in the production of fully dense, shaped parts from prealloyed powders. It has also been used to some extent for the healing of casting defects.

The superalloy powder is packed inside a thin-walled collapsible container, which is a geometrically expanded version of the final shape. After vacuum degassing at an elevated temperature, the container is sealed, pressure checked, and subjected to the simultaneous application of pressure and temperature for a time sufficient to cause full densification of the powder. Finally, the shaped part is obtained by stripping off the container.

At this time, processing of superalloy disks, near-final shape, has become a firmly established technique. The development phase of the process, however, is far from over. Remaining to be formulated are procedures for more effectively controlling the size and distribution of fatigue-limiting flaws, such as ceramic inclusions, which are invariably found mixed in with the alloy powder. The most promising immediate solution to this problem is to screen out the finer size fractions (44 μm) of alloy powder for fabrication purposes. In this way, an upper limit can be set for the maximum flaw (inclusion) size, which controls fatigue crack initiation. This is an important design consideration in highly stressed parts, such as disks.

Hot isostatic pressing has also found useful application in the healing of defects in conventionally cast blades and vanes, such as small shrinkage pores. When such pores are not surface-connected, HIP may be used to close them. Surface-connected pores can be healed only if a coating is applied before HIP. This treatment results in a significant improvement in rupture life.

In addition to these current applications, HIP is also being considered for the rejuvenation of parts, such as blades, vanes, and disks, that have sustained creep or fatigue damage in service. Blade and vane materials that have been extended well into steady-state creep can be restored to their original condition by HIP and reheat treatment. The HIP cycle evidently heals the micropores that develop during creep. Fatigue-induced cracks in disk materials have also been healed by HIP. However, it is not yet clear how best to handle the surface-connected cracks, which tend to be oxidized or otherwise degraded. Another area of interest is the fabrication of laminated composite structures, or graded monolithic structures. A good example of the latter is the dual-property disk concept, which envisages a bore that possesses high load-bearing capacity and a rim that has good creep strength. Conceivably, this structure could be made by HIP, utilizing two different powder compositions.

Diffusion brazing. Conventional brazing is widely used in the fabrication of gas turbine components and parts, but not usually for joining critical superalloy components that are exposed to high temperatures and corrosive environments. This is because it is difficult to design a filler material that satisfies all property requirements, both physical and mechanical, including complete compatibility with the workpiece. In recognition of these shortcomings in conventional brazing, a new process called diffusion brazing was developed in 1974 (*24*).

Diffusion brazing, as applied to the joining of superalloy components, is a vacuum brazing operation. The process makes use of a filler, or interlayer material, that closely matches the composition of the workpiece except for the addition of an appropriate melt depressant, such as boron, to form a eutectic with a low melting point. The interlayer is placed between the mating surfaces of the workpiece, and a slight normal pressure is exerted to maintain physical contact. The temperature is increased to the predetermined brazing temperature, where the eutectic melts and alloys with the workpiece. Under isothermal conditions, the melting point of the eutectic gradually rises as the boron diffuses away into the workpiece. The process is considered to be complete when no eutectic remains in the joint. A subsequent heat treatment is employed to erase all traces of the original interface.

Diffusion brazing produces good joints in a variety of nickel-base superalloys, including dissimilar alloy combinations. Furthermore, bond strengths comparable with base-metal strength have been achieved, even in high-temperature creep-rupture tests. Successful applications for diffusion brazing have included joining of vane clusters, attachment of hardened tips to blades, and bonding of two-part blades (*25*) (Fig. 14). With respect to the two-part blade application, good joints have been obtained irrespective of whether the matching blade halves have DS columnar-grained or single-crystal structures.

Future Perspective

Among the factors that will influence the application of materials to future aircraft gas turbines will be the cost and availability of certain strategic elements, such as Cr, Co, Ta, and Pt. This may dictate the limitation of specific alloys to the most critical engine components. Such constraints will promote the development of new material systems and processes. As design schemes for gas turbine components increase in sophistication, it seems clear that manufacturing innovations will play an increasingly important role.

Some anticipated changes are outlined below:

1) Low- to moderate-temperature static and rotating components will be manufactured from high-performance composite materials.

2) Near-final shape disks will be produced that are graded in composition, microstructure, and anisotropy to optimize mechanical properties.

3) New burner configurations will permit the use of new, improved sheet materials.

4) Multipiece construction of vanes and blades to provide highly efficient cooling will become common practice.

The emphasis in military engines will be on improving system reliability and performance. Higher turbine inlet temperatures will continue to be the primary goal, since this is the most effective way to increase power output. Commercial engines will be designed primarily for fuel efficiency, possibly even at the expense of some sacrifice in component durability. An important consideration will be the attainment of higher metal temperatures for turbine blades, to reduce cooling-air requirements. The achievement of higher design strengths for disks and higher resistance to environmental degradation for vanes will also be important goals, because of the benefits of reduced weight, lower engine cost, and increased operating life.

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Metallic Glasses

John J. Gilman

During almost all of the 8000 years that metals have been used by humans, their structures have consisted of aggregates of crystals. Therefore, the discovery by They showed that very fast cooling ($\sim 10^6$ °C per second) can yield metallic materials that are rigid and have liquid-like molecular structures.

Summary. The novel internal structures of metallic glasses lead to exceptional strength, corrosion resistance, and ease of magnetization. Combined with low manufacturing costs, these properties make glassy ribbons attractive for many applications. These materials also have scientific fascination because their compositions, structures, and properties have unexpected features.

Klement *et al.* (*1*) that selected metal alloys can be quenched fast enough to circumvent crystallization caused considerable exictement among metallurgists.

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By analogy with other supercooled liquids, quenched metallic alloys are called glasses. Since they are derived from liquids rather than gases or plasmas, they do not necessarily have the same structures as other noncrystalline metals. Also, since associations of atoms often exist in liquids, they are not necessarily "amorphous," but instead may contain well-defined short-range order.

Perhaps the prime virtue of metallic glasses is that they can be produced in useful forms economically. As a result, on a comparative cost basis, they are potentially the strongest, toughest, most corrosion-resistant, and most easily magnetized materials known to man. Their costs are very low because they are formed directly from the liquid without passing through the many steps that are used in conventional metallurgy. They can be made from the least expensive of all metallic raw materials, iron.

To make a thin strip of steel in the conventional way, an ingot is first cast; this is hot-rolled to form a billet, the billet is flattened by further rolling into a narrow plate, and then a series of cold-rolling and annealing steps is used to reduce the plate to thin strip stock. In all, six or eight steps are needed.

In contrast, thin strips of metallic glass are cast in one step. An entire spool can be produced in a matter of a few minutes. It is estimated that about four to five times less energy is consumed in going from a liquid metal to a thin, metallic glass strip than would be consumed by

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