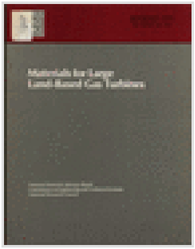

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4/86

MATERIALS FOR LARGE LAND-BASED GAS TURBINES

Report of the Committee on
Materials for Large Land-Based Gas Turbines

NATIONAL MATERIALS ADVISORY BOARD
Commission on Engineering and Technical Systems
National Research Council

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This report has been reviewed by a group other than the authors according to procedures approved by a Report Review Committee consisting of members of the National Academy of Sciences, the National Academy of Engineering, and the Institute of Medicine.

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ABSTRACT

Advanced large land-based gas turbines are expected to be a key component in the generation of electric power based on coal gasification and a combined-cycle gas turbine-steam turbine system. The development of gas turbines in the 120- to 150-MW range with turbine inlet temperatures of 2600°F at pressure ratios up to 16:1 is envisioned over the next 15 to 20 years. Currently available and developing materials technology useful for large machines is reviewed and discussed. Although the primary source of these developments is the aircraft engine field, other sources are also reviewed. Suitable technology appears to be available for the development of higher power, long-life turbines for utility power generation. Recommendations are made for specific research and development efforts that address the special requirements and environment of these machines.

ACKNOWLEDGMENTS

The significant contributions of the committee members and their numerous colleagues whose work appears in this report are gratefully acknowledged. In addition, particular thanks are given to those who made valuable presentations to the committee: Robert Sprague of the General Electric Company and Maurice Gell of United Technologies Corporation gave us reviews of the state-of-the-art materials in aircraft gas turbine engines. George Wacker and Lou Aprigliano of the Naval Ship R&D Center presented an overview of corrosion-resistant alloy development and evaluation. Frank Holden of Battelle Memorial Institute reviewed a study that he led for the Electric Power Research Institute on the causes of failures and shutdowns of gas turbines especially related to materials. Don Moss, plant manager of the Putnam Plant, Florida Power and Light Company, hosted a meeting of the committee and led a very informative tour and discussion with several of his associates on the construction, development, and operation of their excellent gas turbine-steam turbine combined-cycle plant. Carl Lowell presented an overall review of NASA work with emphasis on tungsten wire-reinforced composites and gas path seals, prepared by Robert Signorelli and Robert C. Bill respectively. Louis P. Jahnke provided significant suggestions and comments during the final presentation to EPRI. Wate Bakker and Al Dolbeck of EPRI supplied valuable overviews of the coal gasification combined-cycle plant at the first meeting, hosted the final meeting, and provided strong support and guidance throughout the study.

Finally, special mention and thanks go to Don Groves, who as the professional staff member of the National Materials Advisory Board bore a major share of the responsibility for this committee. His expertise in helping to organize and recruit the committee and on through the efficient setting up and conducting of meetings plus shepherding this final report into being is greatly appreciated. It has been a special privilege to work with Don and to acknowledge his role in the committee's effort.

Louis R. McCreight, Chairman

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SUMMARY

The development and utilization of large land-based gas turbines for generating electrical power have languished since the energy shortages and sharply increased prices of the 1970s. Those events, along with the resulting lower consumption, new lower cost coal, and nuclear capacity that was already under construction and is now on line (or nearly so), have led to very low utilization of gas turbines for power generation in the past decade. With the advent of advanced coal gasification processes such as are being demonstrated on a large scale by the Cool Water Coal Gasification Program at Daggett, California, a new and enlarged role for gas turbines in utility service is anticipated.

Although steady progress toward improved performance has been demonstrated by the combined efforts of both manufacturers and utility operators of gas turbines during the past several decades, it appears that major advances in these machines are both possible and warranted. Several excellent choices of materials and related technologies are available for consideration in future models of gas turbines. They are based on a substantial research and development foundation and in most cases on actual aircraft engine applications. However, there are significant differences between aircraft and land-based gas turbines in the size, rotational speed, life expectancy, time between overhauls, duty cycles, and quality of fuel and air. Some of these factors are more easily accommodated in the land-based gas turbines, and others are easier in the aircraft applications. This leads to a feeling of optimism that the technology transfer can be accomplished as it has been (in both directions) in the past. However, some extensive development, evaluation, and life testing efforts will be needed in most cases.

The materials technology requirements for an advanced land-based gas turbine, operating at a turbine inlet temperature of 2600°F, burning clean, medium-Btu gas, have been considered in conjunction with available aircraft gas turbine technology. It has been concluded that, in general, the requirements for an advanced land-based gas turbine are achievable through the application of materials technology developed or under development by the aircraft engine industry. In certain areas it will be necessary, however, to modify that technology or to develop new technology to

address the special requirements and environment of large land-based gas turbines. Increased difficulties are expected to arise, for example, from requiring more thermal protection by both coatings and air-cooling of the large blades and vanes along with superalloy rotors and more refractory combustion system components. Brief comments and recommendations for these specific areas follow; more detailed conclusions and recommendations are included with each topic in Section 4.

- Compressor--Proposed increases in compression ratio will lead to higher temperatures and stresses in the compressor airfoils. Thus, higher strength materials must be used. Such materials are already in use in aircraft engines. Technology transfer is relatively straightforward and does not require additional R&D effort. Current disk technology is also considered adequate.

- Combustion system--Aircraft combustion technology cannot be transferred directly to land-based engines because of differences in configuration, emission requirements, and fuels.

  Designs to reliably achieve a turbine inlet temperature of 2600°F for long periods of time need to be developed. It is anticipated that such designs will require advanced materials and cooling technologies. Since combustion system components often have the shortest life in the turbine, early attention should be given to selecting and developing the most promising materials technologies. Potentially applicable technologies to be investigated more closely include advanced thermal barrier coatings, ceramic composites, oxide-dispersion-strengthened alloys, and fabrication techniques allowing improved air cooling.

- Turbine airfoils--Proposed increases in turbine inlet temperature to 2600°F will require more advanced cooling as well as higher strength superalloys capable of operating at higher temperatures, for both stationary and rotating vanes. In aircraft turbines, inlet temperatures in excess of 2600°F have been achieved by using directionally solidified and single crystal airfoils together with advanced cooling and coatings.

  Casting technology should be developed for producing directionally solidified and single crystal airfoils in the sizes required for large land-based gas turbines. To assess the integrity of these large castings, the mechanical properties of cast components should be evaluated by conducting appropriate mechanical property tests on specimens that are extracted from selected airfoil and attachment locations.

  Protective coatings will be required to provide oxidation resistance, protection against environmental embrittlement, and occasional hot corrosion resulting from possible upsets in the gas clean-up system or inlet filtration. Coating compositions are generally available from the aircraft industry. However, their ability to survive the much longer service lives required in land-based gas turbines must be established.

  Thermal barrier coatings on stationary and rotating airfoils will be extremely helpful in increasing service life, reducing thermal stresses, and increasing turbine efficiency through reduced cooling requirements. A significant effort is therefore recommended to develop long-life thermal barrier coatings for this application. Special emphasis should be given to oxidation of bond coats, corrosion resistance, thermal cycling resistance, and long-term testing.

Advanced fabrication technologies (e.g., spar-shell or joining separate airfoils and shroud castings) can provide an alternate way of introducing the degree of cooling required for first-stage vanes. Work in these areas may be required if scale-up of aircraft directionally solidified and single crystal castings proves impractical.

Alloys and cooling technology currently used in the first-stage airfoils of land-based gas turbines must be transferred to the later stages of the 2600°F turbine. This may require development efforts in casting technology, forging, and refinement of casting alloys because of the large size of these components. The manufacture of the last-stage blades will be especially challenging.

- Turbine disks--The proposed increases in turbine inlet temperature to 2600°F will require disk alloys capable of operating at higher stresses and temperatures. Nickel alloys will most likely be required. Properties of such alloys must be determined. The producibility of large disks should be reaffirmed. Nondestructive tests for large disks must also be established.

- Turbine seals--The efficiency of a gas turbine can be significantly improved by optimizing various seals and clearances throughout the engine. The high-pressure, high-temperature turbine blade outer air seal is especially important. For aircraft turbines a 0.5 percent reduction in fuel consumption can be obtained for every 0.25-mm (0.01-in.) reduction in operating tip clearance. Technology recently developed for aircraft (e.g., abradable seals) should be modified to obtain the longer life required for land-based engines. A developmental thick thermal barrier coating type of seal should be investigated to assess its use in land-based gas turbines.

- Life prediction--Life prediction methods and supporting data for the various combinations of environments, times, temperatures, and duty cycles predicted for future large land-based gas turbines, especially their hot gas path components, are not well developed and need to be extended. Programs to collect appropriate data and to extend life prediction methodology focused on land-based gas turbine service requirements should be supported. Coordination with nondestructive evaluation techniques and advances is necessary, especially for turbine disks, blades, and vanes.

- Federal agency program information--There are extensive research and development programs supported by DOD, DOE, and NASA of interest to the development of large land-based gas turbines for which the reports distribution is carefully controlled. It is recommended that EPRI and all other interested parties establish ties to obtain these reports and related information on a continuing basis.

- Component field-testing--It is strongly recommended that plans be developed for component field-testing to evaluate materials and processes. In many cases components can be produced that are suitable for near-term evaluation of advanced concepts in currently operating engines. Through its contacts with the utility industry, EPRI can be particularly helpful in these programs by arranging for such tests.

Section 1

COMPARISON OF AIRCRAFT AND LAND-BASED GAS TURBINES

Because of the importance of aircraft gas turbines for military and commercial applications, a large research and development program directed toward improving their performance and durability has been carried out for more than 40 years. As a result of this effort, the thrust efficiency (i.e., the thrust power output divided by the heat energy rate provided by the fuel) has increased from 22 percent in 1958, the date of first commercial service, to over 35 percent in 1984. The modern aircraft gas turbine represents the frontier of technology in many areas of materials, coatings, heat transfer, aerodynamics, structural design, and other important technical disciplines.

The large land-based gas turbine has not benefited directly from military funding, nor does it enjoy the same commercial environment that has stimulated the aircraft gas turbine technology. In the early 1970s a significant number of gas turbines were purchased by public utility companies for use in generating peak power, and a limited number were purchased for base-load applications. As a peak-load power plant, the gas turbine enjoyed the advantages of low capital cost and short-time delivery, installation, and start-up. As a base-load power plant, the gas turbine in combination with a steam turbine that utilized the exhaust heat offered a higher overall efficiency than a conventional steam plant. However, the oil crisis of 1973 and the associated rise in the cost of distillate fuels greatly increased the operating cost of gas turbine power plants. Today gas turbines are generating only about 1 percent of the power in the United States. As a result of the current small market for large land-based gas turbines, little funding has been available for advancing the state of the art of this type of power plant.

The Electric Power Research Institute (EPRI) foresees, however, a new and important role for the land-based gas turbine as a result of the development of coal gasification technology. In an integrated coal gasification combined cycle (CGCC) plant, electricity is generated from coal with a relatively high efficiency and extremely low SO_2 and NO_x emissions, as compared to conventional coal-fired power plants. Figure 1-1 is a block diagram of a typical CGCC power plant using an oxygen-blown entrained slagging gasifier. The coal is gasified with oxygen and steam to produce

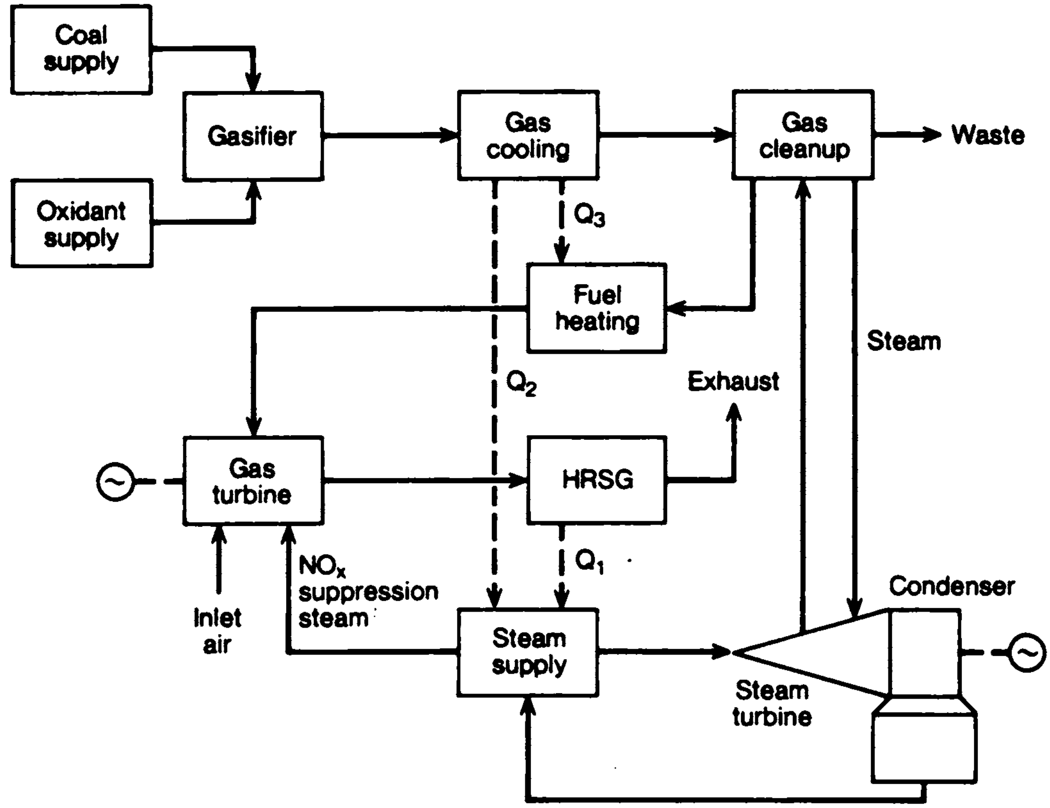


Figure 1-1. Block diagram of oxygen-blown coal gasification combined cycle power plant.

a raw syngas consisting mainly of CO and H_2 but with substantial quantities of CO_2 and H_2O, minor quantities of H_2S, NH_3, and HCl (200-10,000 ppm), and a few parts per million of alkalis. After gasification, the raw syngas is generally cooled in radiant and/or convective boilers, where most of the slag is separated from the gas. Following heat recovery and cooling, the gas is quenched in a scrubber, where the remainder of the particulates are removed with most of the NH_3, HCl, and alkalis. The partially cleaned syngas is further cooled down and fed into a sulfur removal system, where the H_2S is removed, usually with an organic solvent. This process is 95 to 99 percent effective. Following the sulfur removal, the clean gas is scrubbed once more to remove traces of organic solvent. Before entering the gas turbine, the dry clean syngas is saturated with boiler feed water and reheated to 300-400°F. At this point, a typical gas composition would be 35 percent CO, 27 percent H_2, 13 percent CO_2, 24 percent H_2O, 0.04 percent CH_4, 0.4 percent Ar, 1.2 percent N_2, 4 ppm H_2S, 60 ppm (max.) COS, 10 ppm (max.) HCl, alkalis not detectable. Thus the medium-Btu gas will have a purity comparable to natural gas and high-quality liquid jet engine fuels.

The CGCC power plant will compete with conventional advanced coal-burning power plants such as fluidized-bed boilers and with nuclear power plants for its share of the U.S. electric power generation market. Decisions on which type of power plant to buy will be made largely on the projected cost of electricity of each system. The efficiency and power output of the gas turbine and its high exhaust temperature are major factors for determining the overall efficiency of a CGCC plant. Thus, the economic success of CGCC technology is in part dependent on the development of improved land-based gas turbines with high efficiency and capacity.

The high technology of the aircraft gas turbine can make a contribution to the land-based gas turbine in two important areas: the level of the burner discharge temperature and the efficiency of the compressor and turbine components. Current commercial aircraft gas turbines operate at a maximum burner discharge temperature of approximately 2800°F, whereas land-based gas turbines operate at a maximum temperature of about 2200°F. The component efficiencies cannot be compared directly because the aircraft parts are somewhat compromised to achieve light weight. In a proper comparison, however, the aircraft components would show an advantage of several percentage points. Some of this aircraft technology is being incorporated in the latest models of land-based gas turbines on a continuing basis, but, to take maximum advantage of this advanced technology, new land-based gas turbine designs would be required.

A controlling element in the transfer of aircraft gas turbine technology to land-based gas turbines is the difference in operating requirements and the difference in operating environment between the two power-plant types. These differences are discussed in more detail below as they would apply to a gas turbine operating in combination with a coal gasification plant. For the purposes of this report such an advanced land-based gas turbine is assumed to have the following design characteristics:

Power: 100 to 150 MW, not including steam cycle

Burner Discharge Temperature: 2600°F

Pressure Ratio: 16/1

Fuel: The aircraft gas turbine burns a clean liquid distillate fuel (Jet A). The land-based gas turbine would burn the discharge from the coal gasification system previously described. This gas is quite free of contaminants (with the possible exception of occasional upsets in the gas clean-up operations) that could damage the critical high-temperature turbine parts. It should therefore be possible to directly transfer aircraft turbine technology. Some development work would be required, but no major obstacles are foreseen.

Air: The combustion air at land-based gas turbine sites can be contaminated with salt and other pollutants. It is believed that the use of a high-performance inlet filter can clean up the air sufficiently to permit the use of aircraft high-temperature technology. Usually it is necessary to add steam to the air to meet the NO_x requirements for an industrial power plant. The presence of steam may, however, cause materials and coating problems not encountered in aircraft service. It is important to realize that even a brief exposure (less than 24 hours) to fuel or air contaminated with sulfur, sodium, potassium, chlorine, or other compounds can initiate hot corrosion that may continue after fuel and air purity has been restored. Continuous monitoring and the provision of protective coatings is therefore advisable.

Type of Operation: The aircraft gas turbine is subjected to more cyclic power operation than the land-based gas turbine. A typical aircraft power plant might encounter 15,000 to 30,000 cycles of idle to full power in its operating life, whereas the land-based gas turbine in a CGCC plant is projected to experience only 1000 to 5000 cycles. The aircraft gas turbine is accelerated and decelerated quickly, while the land-based turbine can usually be programmed to change power slowly to minimize thermal loads. In the event of a trip, however, the land-based turbine will be subjected to a very rapid cooling cycle.

Durability: The aircraft gas turbine is generally considered to have a design life of the order of 30,000 hours, whereas the industrial gas turbine has a projected life of 100,000 to 200,000 hours. Many industrial gas turbines have indeed achieved such lifetimes but at lower temperatures than contemplated for advanced turbines.

Maintenance: Major maintenance and repair of an aircraft gas turbine is done in an overhaul shop. It is generally not economical to remove a large land-based gas turbine from a site and return it to the factory. Thus, although major components may sometimes be removed and overhauled at the factory, provision must be made for component inspection and repair or replacement at the site.

Emissions: The land-based gas turbine must meet a different set of emissions requirements than the aircraft turbine. This may prevent the direct application of aircraft combustion technology. With the clean gaseous fuel from the coal gasification process, discharge of smoke, sulfur compounds, or unburned hydrocarbons should not be a problem. However, the high pressure and temperature of the advanced gas turbine cycle will increase the formation of nitrogen oxides. Unique designs such as catalytic combustors or rich-lean combustors may be required to meet emission requirements.

Economics: The most important constraint on the transfer of aircraft technology to land-based gas turbines may turn out to be economic rather than technical. The trade between cost and performance will have to be carefully evaluated when making design choices for an advanced land-based gas turbine. There are compelling reasons to believe that a gas turbine, properly designed and operated, can be expected to match or exceed the performance of a steam plant with regard to durability and maintenance cost.

At present, operating experience with aircraft gas turbines is accumulating at a rate of approximately a million hours a week. Data from this vast information bank are continually analyzed and fed back into the design system. Many individual engines are reaching 20,000 to 30,000 hours of operating time. Such performance presages a life of over 100,000 hours for advanced industrial gas turbines in base-load operation, since the industrial gas turbine will be operated at a lower burner discharge temperature than the aircraft turbine. A reduction of 100 to 150°F can result in an increase of 2 to 4 times in the life of critical hot-section parts.

In summary, the large land-based gas turbine should be capable of long life for base-load operation in combination with a coal gasification plant. Hot-section corrosion of the type experienced on earlier gas turbines can be minimized. Material development and design selection should be related to an anticipated environment of clean fuel, clean air, and burner discharge temperatures 200 to 400°F higher than current state-of-the-art industrial practice.

Section 2

AERONAUTICAL PROPULSION MATERIALS PERSPECTIVE

This section provides a brief perspective on aircraft engine materials to supplement the previous comparison of aircraft and land-based gas turbines and to complement the more detailed following section on materials for land-based gas turbines. This section, rather than being as extensive as the discussion of materials in the next section on land-based turbines, is deliberately brief for two reasons. First, the detailed presentation of advanced gas turbine technology is subject to export control limitations; second, the more pertinent discussion of the committee's recommendations for materials and processes applicable to advanced land-based gas turbines is contained in Section 4. Because much but not all of the matter presented in Section 4 is based on the aircraft turbine engine field, it would be repetitious to also discuss it here.

Since the introduction of gas turbines for aeronautical propulsion, the demands for higher temperatures to improve performance have led to advances in the capabilities of hot-section materials. From the 1950s to the present, these advances have been primarily through the introduction of and subsequent improvements in nickel- and cobalt-based superalloys. Figure 2-1 shows the trends in materials over the years and the potential of future materials.

In the 1950s, most superalloys were used in the wrought condition, which limited the extent of alloying possible. Improvements in casting technology, allowing increased reliability, led to the introduction of materials with improved strength at even higher temperatures through the 1960s. In the 1970s, advances in casting led first to directionally solidified and ultimately to single-crystal nickel-based superalloys, which at present are bill-of-materials for first-stage blades on current, advanced gas turbines.

Figure 2-2 shows the current requirements for materials in turbofan engines. Note that, as the strength-limiting temperatures of materials have advanced to 1900 to 2000°F, oxidation and hot corrosion have become increasing factors in limiting life. Currently, the materials in the hottest stages of such turbines must be coated to protect them from the environment. However, coating technology has kept

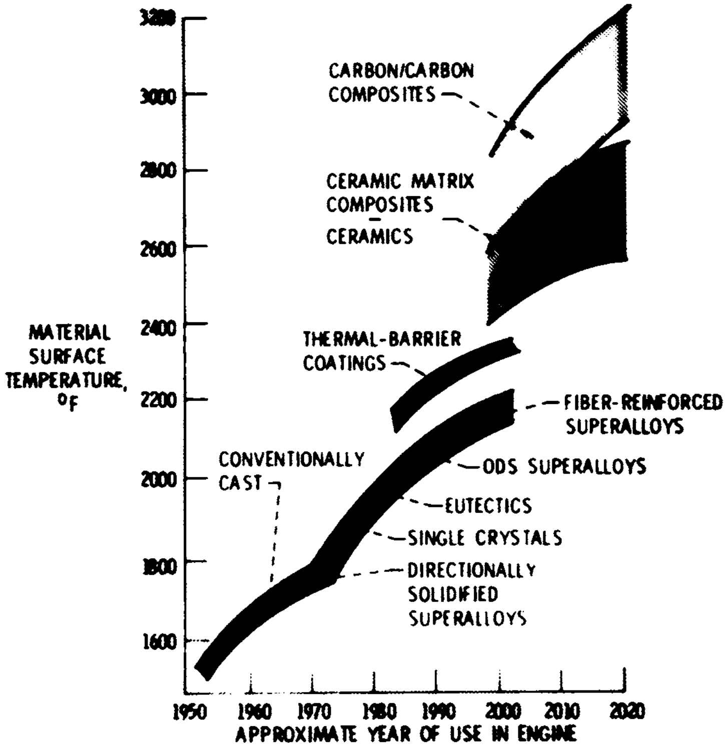
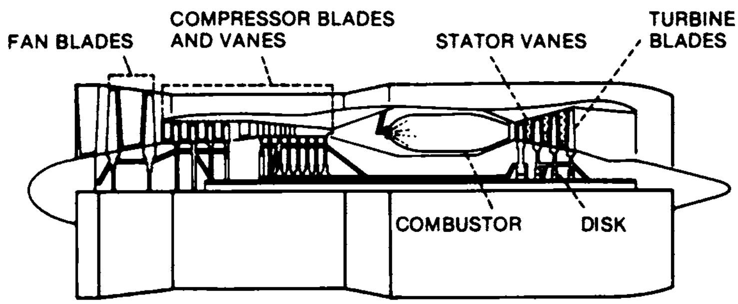


Figure 2-1. Blade material temperature trends in gas turbines for aeronautical propulsion.



COMPONENT	TYPICAL OPERATING CONDITIONS			CRITICAL PROBLEMS
	TEMP (°F)	STRESS (ksi)	LIFE (hr)	
BLADES	1700-1900	30-20	5000	CREEP STRENGTH, STABILITY, OXIDATION, HOT CORROSION, THERMAL FATIGUE
VANES	1800-2000	10-5	5000	THERMAL FATIGUE, OXIDATION, HOT CORROSION
DISKS	800-1200	150-80	15000	LOW CYCLE FATIGUE
COMBUSTORS	1600-2000	5-3	4000	THERMAL FATIGUE, OXIDATION

Figure 2-2. Current typical material requirements in turbofan engines.

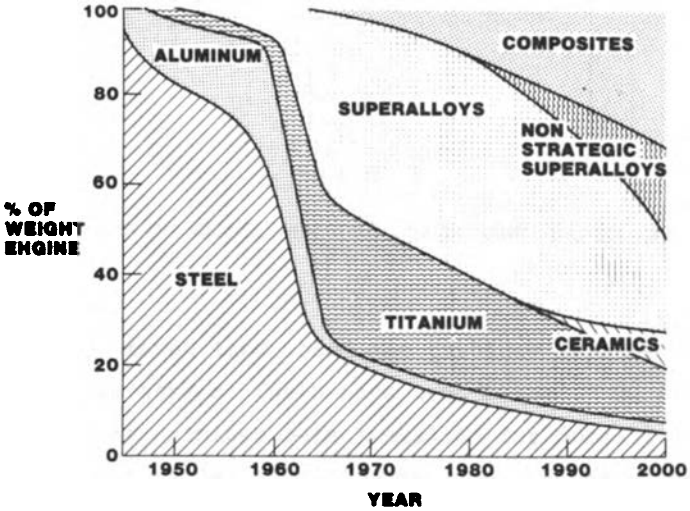


Figure 2-3. Past and projected trends in types of materials for aeronautical propulsion.

pace with alloy development through the use first of diffusion bonded aluminides and subsequently of overlay coatings of the MCrAlY types, where M is nickel or a combination of nickel and cobalt. More recently, the introduction of thermal barrier coatings (a partially stabilized zirconia over a MCrAlY bond coat) has offered the potential of even greater protection and perhaps higher turbine inlet temperatures.

Materials improvements have been matched by the introduction and refinement of airfoil cooling technology to allow even higher turbine inlet temperatures. Even though the use of advanced film cooling has made this technology extremely efficient, there is still a substantial penalty in specific fuel consumption. The twin advances in alloys and cooling are nearing the ultimate in potential turbine temperature, as can be seen in Figure 2-1, and future improvements will probably require the introduction of new classes of materials (e.g., composites) or coatings (e.g., thermal barrier coatings, which are currently being introduced on vane platforms in some advanced engines).

The use of new material systems (alloys plus coatings) offers the promise of even higher temperatures in the future. Figure 2-3 shows one projection of the introduction and use of new materials. The exact trend lines are conjectural at best, but it is clear that future needs will demand increased use of high-temperature composites of all types--polymer matrix, metal matrix, ceramic matrix, and perhaps even carbon matrix. How rapidly these materials and their coatings will be introduced will in part depend on the extent of the resources applied to their development.

Section 3

STATE OF THE ART OF MATERIALS FOR LARGE LAND-BASED GAS TURBINES

During the past 30 years, large land-based gas turbines have been used in numerous applications in which a variety of fuels have been burned. During the late 1960s and 1970s a large number of machines were installed for domestic electric utility peaking service involving very few operating hours per machine start. Thus these machines have accumulated relatively low unit operating hours compared to a base-load mode of operation.

Experience in base-load operation (i.e., continuous running) has been accumulated mainly in the industrial market for pipeline and industrial process drives as well as industrial power generation. Internationally, the gas turbine market for base-load utility power generation has shown a steady growth during the past 10 to 15 years. More recently, interest has increased domestically for base-load units with the introduction of combined-cycle plants and the emergence of the cogeneration market.

Service hours for the industrial base-load machines are accumulated rapidly, with many machines of older design, such as the MS-5001, MS-3002, W101, and W191, experiencing over 100,000 hours of operation, and a fair number having operated for over 200,000 hours. In more recent designs, such as the MS-7001 and W501, fired hours have been expectedly lower, although many have operated for over 50,000 hours, with several in service for over 100,000 hours.

With the advent of increased firing temperature in the late 1960s and 1970s, machines burning liquid fuels often encountered turbine blade hot corrosion problems as a result of contamination of the fuel during transportation and also from contaminated inlet air. These problems were greatly alleviated with the introduction of more corrosion-resistant turbine blade alloys, the use of corrosion-resistant blade coatings, and the use of inlet air filtration. It is interesting to note that the base-load service experience cited above involved primarily the use of clean gaseous fuels. Experience with such fuels has been excellent relative to the occurrence of hot corrosion, a point believed pertinent to this study.

The growth in use of large land-based gas turbines has been the result of improved designs involving increased air flow, firing temperature, and equipment size. Advances in compressor development during the past 20 years consist of increases in compressor pressure ratio and air flow, as shown in Figures 3-1 and 3-2.

The increase in turbine firing temperature since 1950, defined as the gas temperature at the exit of the first-stage guide vane, is shown in Figure 3-3. The increase in firing temperature has been the result both of design improvements and of the development of hot-gas-path materials with increased temperature capability. Design improvements include the use of lower stress hollow blades, control of gas-path radial temperature profile, and, starting in about 1960, the use of vane and blade air cooling. Materials developments include the use of vacuum-processed nickel-based superalloys, improved cobalt-base alloys, and coatings resistant to oxidation and hot corrosion. Many of the materials developments have been the result of technology transfer from the aircraft gas turbine industry. However, because of large differences in component size, as shown in Figure 3-4, and because of the more variable fuel and air quality encountered, unique materials and processing requirements have confronted industrial gas turbine materials engineers.

Current belief is that the industrial gas turbine will play an increasing role in future base-load power generation in the form of a combined gas-steam plant burning

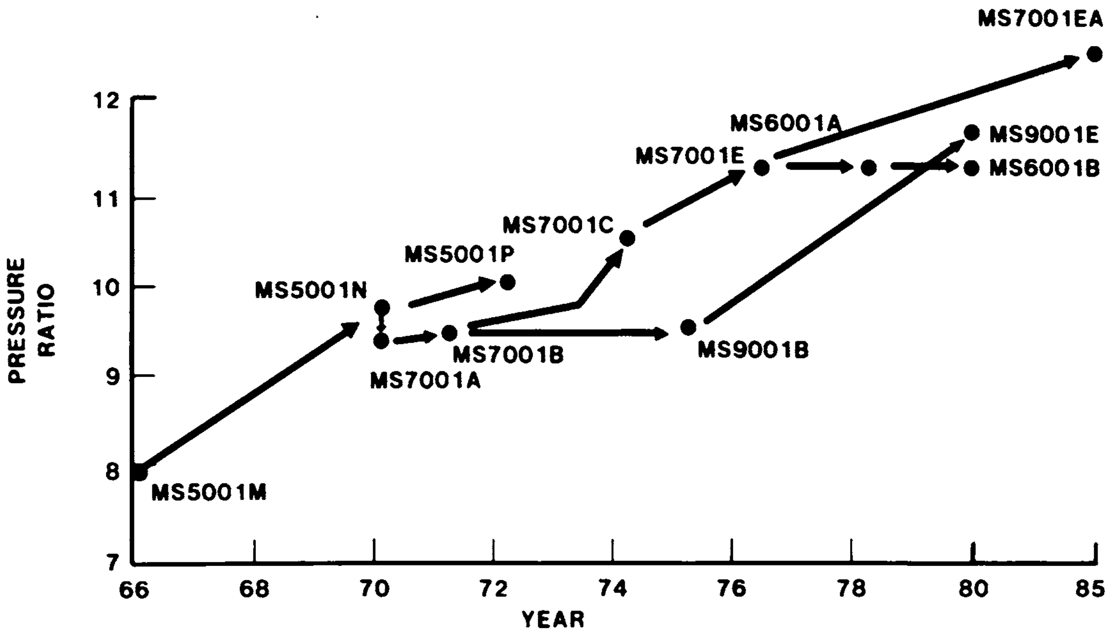


Figure 3-1. Growth in compressor pressure ratio (International Standards Organization conditions) (1).

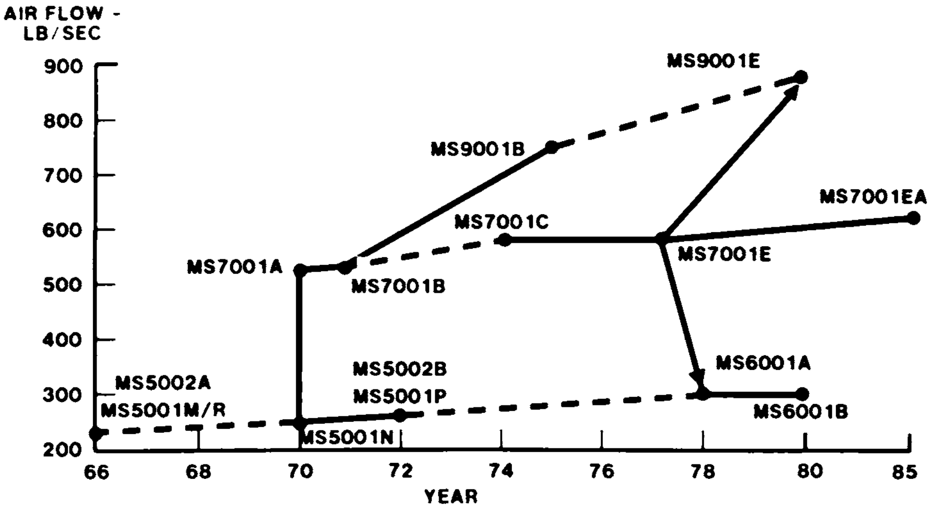


Figure 3-2. Growth in compressor air flow (International Standards Organization conditions) (2).

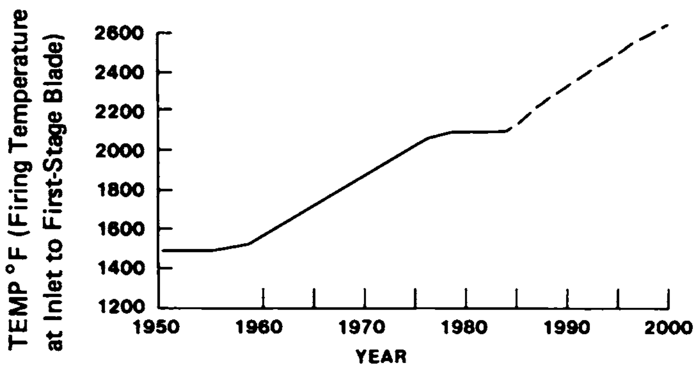


Figure 3-3. Past and future trends of GE heavy-duty gas turbine firing temperatures (3).

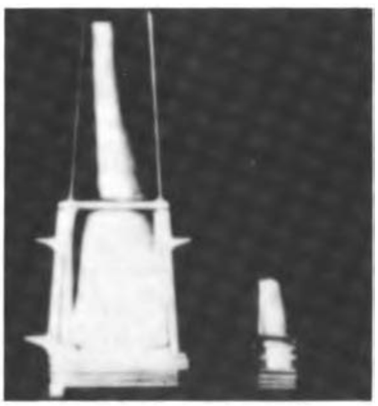


Figure 3-4. GE land-based MS7000 and aircraft engine CF6 blades.

clean gasified coal. Because of the projected use of clean gas, hot corrosion is not expected to pose a limitation on component longevity, a point that has been well demonstrated by experience with turbines operated with natural gas. It is expected that such plants will operate at even higher firing temperatures than used today for increased specific output and thermal efficiency. The value of higher firing temperatures is shown in Figures 3-5 and 3-6, where the projected increase in output and thermal efficiency is shown as a function of firing temperature and is compared to current turbines.

It is believed that a significant contribution to these advances will be achieved through improved materials and associated processing. As a reference for discussion of fruitful areas of materials and processing development, this section primarily reviews the current state of the art of U.S. industrial gas turbine materials along with current judgments of future changes. Two common U.S. land-based gas turbine designs are shown in Figures 3-7 and 3-8. A brief review of some European and Japanese programs on advanced industrial gas turbines is then presented at the end of the section.

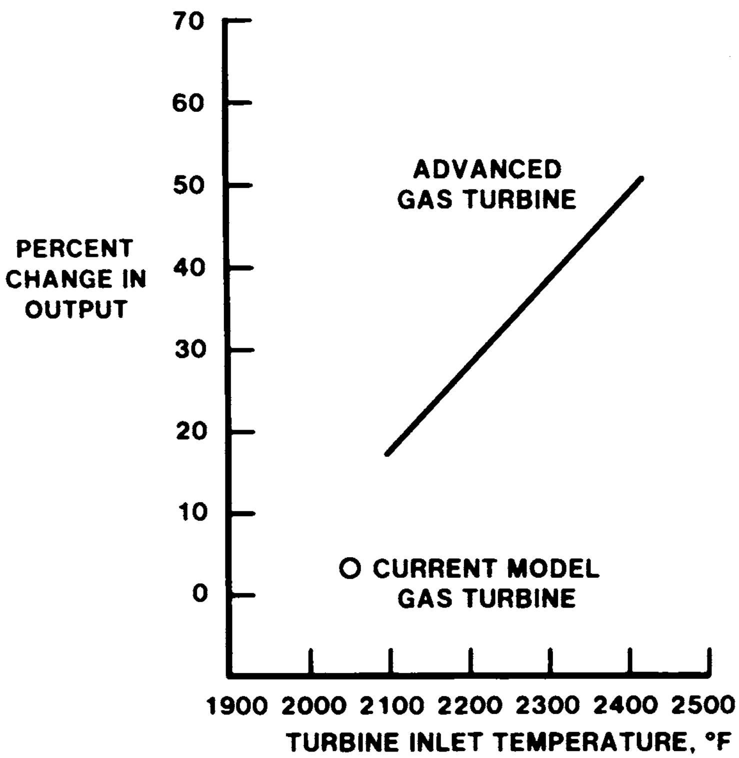


Figure 3-5. Projected increase in output for advanced gas turbine (air-cooled stage-1 vane, entrained flow gasifier, oxygen-blown, using Illinois No. 6 coal).

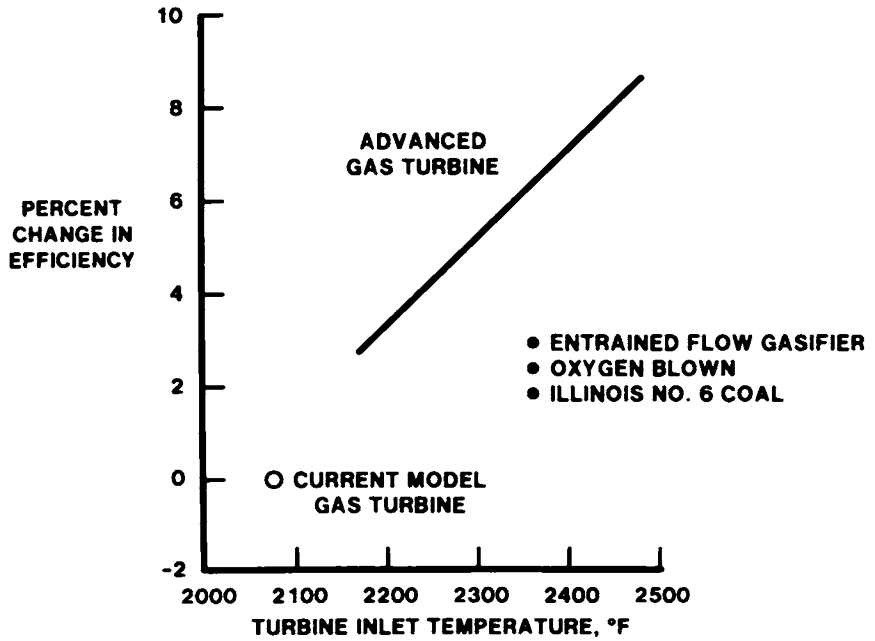


Figure 3-6. Projected increase in thermal efficiency for advanced gas turbine (air-cooled stage-1 vane, entrained flow gasifier, oxygen-blown, using Illinois No. 6 coal).

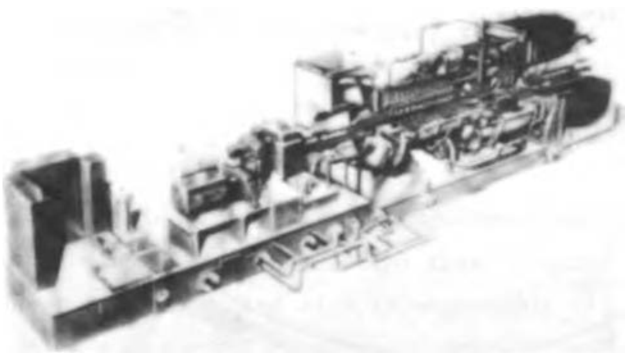


Figure 3-7. MS7001 simple-cycle, single-shaft, heavy-duty gas turbine. (Source: General Electric Company)

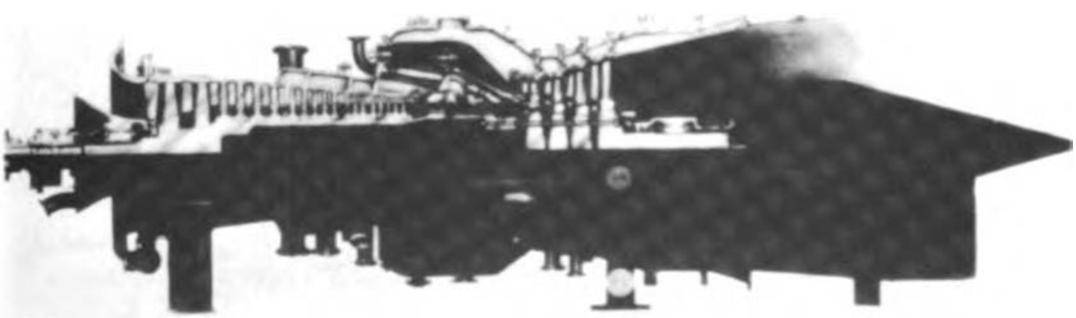


Figure 3-8. W-501D gas turbine. (Source: Westinghouse Electric Corporation)

COMPRESSOR AIRFOILS

Compressor airfoils (stationary vanes and rotating blades) are generally produced from a heat-treatable 12 percent chromium stainless steel such as Type 403 or 410. This class of alloy provides the tensile strength at ambient temperatures and the strength retention and creep resistance needed for the entire temperature range of compressor vane service (up to about 750°F). Other qualities of this class of alloy that are attractive for the compressor blading applications include toughness and resistance to impact caused by foreign objects, good mechanical damping, and resistance to atmospheric corrosion in a variety of environments.

Rotating blades are either precision-forged or machined with very thin leading and trailing edges for aerodynamic efficiency. The stationary vanes, like blades, may be produced as individual pieces by forging, extrusion, machining, or rolling. In addition, vanes may be assembled into multiple segments of up to 180 degrees. In the latter case the vanes are assembled into segments and welded to curved shroud strips. The good weldability of the 403 stainless steel is a property that is central to this manufacturing technique. Some compressor vanes are produced as precision castings containing several airfoils in a single segment of approximately 30 to 60 degrees (Figure 3-9).

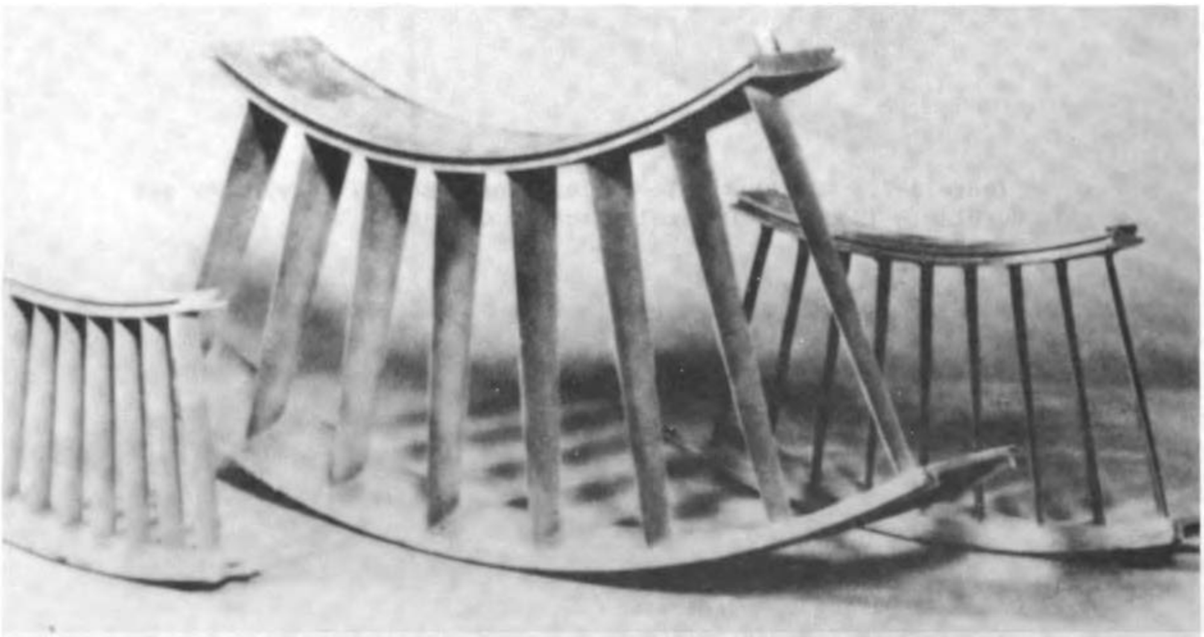


Figure 3-9. Cast 12 percent chromium steel compressor vane segment.

Corrosion of the compressor vanes is generally not a problem unless the turbine is operated in salty or acidic environments or intermittently in humid climates. Corrosion of blades may also result from periods of extended storage without adequate protection. In these cases corrosion has been observed. Compressor efficiency is significantly deteriorated by vane surface roughness. However, a protective coating can be applied to the blades and vanes to maintain performance. Nickel-cadmium plating has a history of successful resistance to compressor vane corrosion in marine environments. Aluminum-rich sacrificial coatings have also proved effective, particularly in acid environments. They may be applied either by a sprayed slurry followed by low-temperature baking or by a diffusion process. In the second case, care must be taken to ensure that the coating application temperatures do not adversely affect the temper condition of the base alloy.

For compressors in state-of-the-art land-based gas turbines, use of a coating on type 403 or 410 stainless steel has significant cost benefits over the use of a corrosion-resistant nickel alloy or titanium. However, these latter materials, which are common in aircraft turbine engines, may be necessary in future generations of land-based turbines with high compression ratios. In particular, nickel-based alloys may be required for latter stages where compressed air temperatures would exceed 750 or 800°F. A transfer of the necessary technology is not expected to pose any significant difficulties.

COMBUSTION SYSTEM

The combustion system in an industrial gas turbine consists basically of a combustion chamber or chambers and transition ducting that directs the hot gases to the first-stage guide vanes of the turbine with an acceptable temperature pattern. In the most popular design, a number of combustors (cans or liners) and transition ducts are arranged in an annular fashion around the axis of the turbine. A mixture of compressor discharge air and fuel is burned near the neck. The combustor is designed to contain the flame, mix in diluent air for the control of temperature, emissions, and smoke, and provide for air-cooling of the metal walls, as shown in Figure 3-10. A similar design is shown in Figure 3-11.

For this type of service, materials chosen must possess high-temperature strength, including tensile and creep strength, as well as resistance to high cycle fatigue, low cycle fatigue, oxidation, and carburization. Further, the materials should maintain metallurgical stability in service to avoid embrittlement and be fabricable and weldable in sheet form, both for initial manufacture and for ease of repairing service-induced defects.

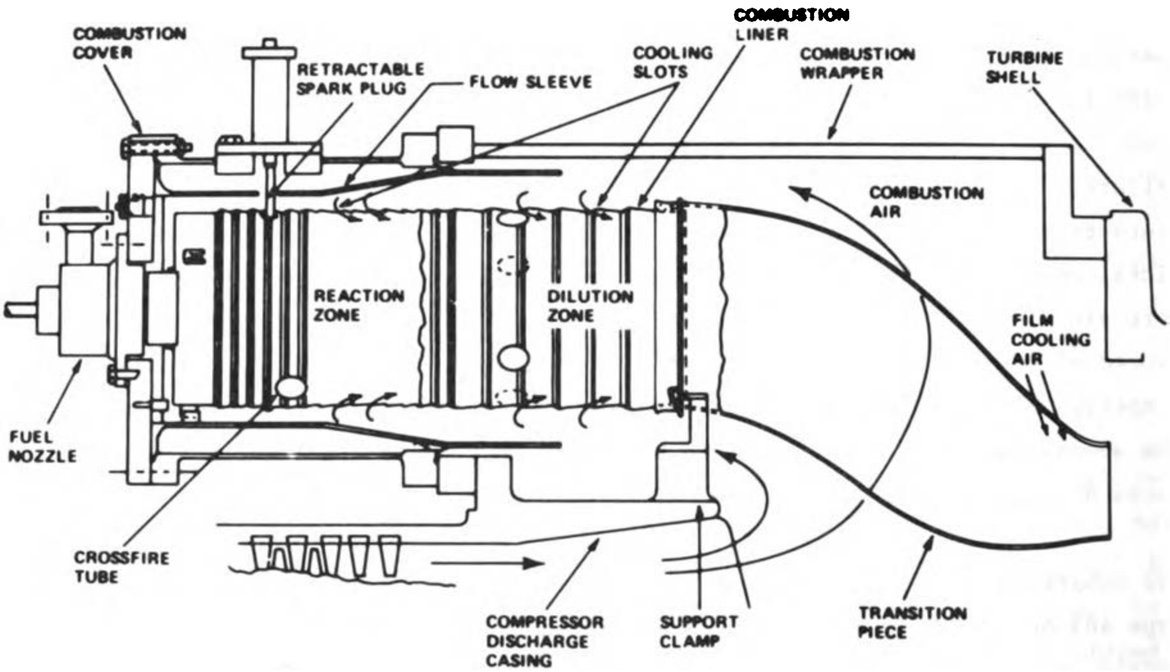


Figure 3-10. MS7001 B/C, C, E, EA reverse-flow combustion system. (Source: General Electric Company)

An adequate cooling design is essential to prevent excessive radiant heating of the combustor walls. If this is done and regular fuel nozzle maintenance is performed to prevent flame impingement, several years of satisfactory service are obtained from the combustor material.

Typical alloys used for industrial gas turbine combustors and transition pieces are listed in Table 3-1. The austenitic grades of stainless steel (e.g., Type 310) were initially chosen for combustors and are still being used successfully. Turbines with higher firing temperatures have generally employed nickel or cobalt alloy sheet, Hastelloy X and Haynes Alloy 188 being popular choices. For improved creep strength retention, alloys such as Inconel 617 and 625 and Nimonic C-263 are being used for some of the newer designs. Nominal stress rupture properties for selected

Figure 3-11. Combustor and transition duct, Hastelloy X.

Table 3-1

SUPERALLOY SHEET MATERIALS USED FOR INDUSTRIAL
GAS TURBINE COMBUSTORS AND TRANSITIONS

Alloy	Ni	Fe	Co	Cr	Mo	W	Al	Ti	C	Other
Nickel base										
Hastelloy X	Bal	18.5	1.5	22.0	9.0	0.6	--	--	0.10	--
Inconel 617	Bal	1.5	12.5	22.0	9.0	--	1.2	0.3	0.10	0.2 Cu
Inconel 625	Bal	2.5	--	21.5	9.0	--	0.2	0.2	0.05	3.6 Cb
Nimonic C-263	Bal	--	20.5	20.0	5.9	--	0.5	2.2	0.06	--
Cobalt base										
Haynes Alloy 188	22.0	1.5	Bal	22.0	--	14.0	--	--	0.1	0.07 La
Iron base										
AISI Type 310	20.0	Bal	--	25.0	--	--	--	--	0.1	--

sheet alloys are shown in Figure 3-12. Whereas Hastelloy X is usually used in the solution-annealed condition, some of the higher strength alloys require additional aging heat treatments to develop their full strength potential. Care must be exercised not to use alloys that, because of their complex physical metallurgy, are difficult to weld or are unstable in service. When the latter condition occurs, service aging can result in the formation of an embrittled microstructure.

Fabrication of combustors is normally by cold-forming of sheet, TIG welding, and spot welding. In some cases brazing is also used. Dimensional accuracy is especially important so that proper alignment can be achieved upon installation.

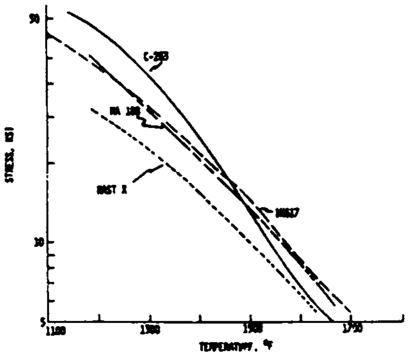


Figure 3-12. Stress-rupture properties of combustor sheet alloys after 1000 hours.

Transition ducts are fabricated from the same class of alloys (and often the same alloy) as the combustors. The property requirements are quite similar, as are the fabrication methods.

Coatings are sometimes applied to combustors and transition ducts. Aluminides or other diffusion-type coatings can be used for protection against corrosion and oxidation. Ceramic thermal barrier coatings, such as stabilized zirconia, have been used to reduce metal wall heating due to radiation. These coatings are applied by plasma spray and may consist of at least two layers; a thin sprayed metallic bond coat (0.005 in.) to metallurgically adhere to the basket inner wall and a thicker (0.010 to 0.015 in.) ceramic insulating layer (Figure 3-13). Thermal barrier coatings have shown metal temperature reductions in combustion liners of 150°F and have excellent potential for higher temperature combustion turbines.

Experience indicates that the combustion system is often the component requiring most frequent inspection and maintenance or replacement. While the design of large land-based gas turbines incorporates features for ease of combustor maintenance, clearly some introduction of advanced technology that would allow use in higher temperature machines with extended inspection, repair, or replacement intervals would be desirable. Unfortunately, operating requirements such as control of NO_x and other emissions have not been well defined. The eventual solutions (e.g., steam or water injection, innovative design, staged combustion, catalytic burner) will greatly influence the extent to which advanced aircraft engine technology can be adapted to large land-based gas turbines.

Advanced oxide-dispersion-strengthened (ODS) materials now in limited use in some aircraft combustion systems may be introduced in future industrial gas turbines.

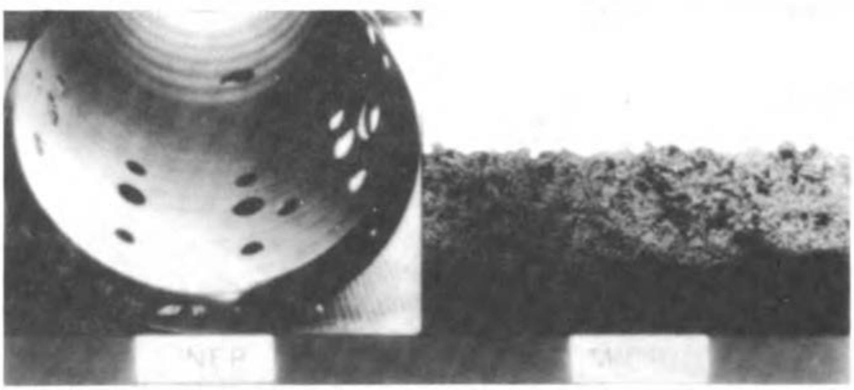


Figure 3-13. Thermal barrier coating applied to combustion liner (3).

Figure 3-14. Schematic diagram of Lamilloy.

The high-temperature creep resistance of ODS alloys is very attractive, but they present problems for fabrication by conventional fusion welding techniques because of the agglomeration of oxides in the overheated zone and the accompanying strength loss. Diffusion bonding, brazing, and mechanical joining are, however, possibilities.

Advancements in cooling technology, such as the use of Lamilloy (a trademark of General Motors Corporation) or other quasi-transpiration cooling schemes, represent another way of extending the temperature capability of combustor materials without sacrificing service life. Lamilloy is a method of construction, shown schematically in Figure 3-14, that consists of either a single perforated sheet or a sandwich of bonded sheets separated by pedestals with small holes in the sheets. The holes and pedestals are photochemically etched in the sheet surface prior to diffusion bonding. The combined pedestals and holes form channels through which cooling air passes, providing a combination of impingement and transpiration cooling. When used for combustor basket fabrication, the cooling air passes from the outer surface to the inner. The Lamilloy combustor shown in Figure 3-15 was cooled very uniformly in a series of tests.

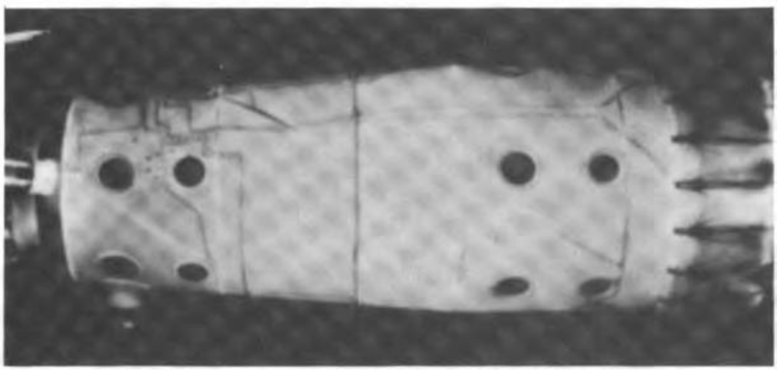


Figure 3-15. Lamilloy combustor.

TURBINE VANES

The first-stage turbine inlet guide vanes (also sometimes called nozzle partitions) must perform the function of turning and directing the flow of hot gas into the rotating stage of the turbine at the most favorable angle of incidence. They are literally right in the line of fire, being routinely subjected to impingement of the highest temperature gases and experiencing the highest metal temperatures of any component in the turbine. Even though the superalloys used for vanes are capable of creep resistance at temperatures above 1700°F for short periods in aircraft engine applications, the desire for component lifetimes of 50,000 to 100,000 hours for industrial turbine vanes means that a high degree of cooling is necessary. Representative designs are shown in Figures 3-16 and 3-17.

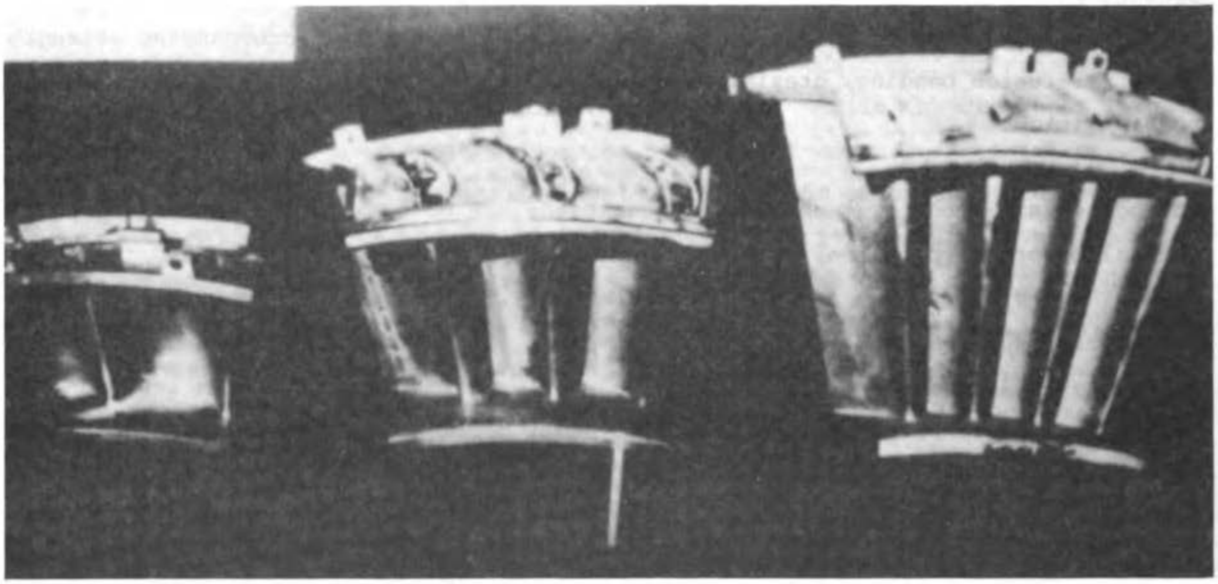


Figure 3-16. Turbine vanes.

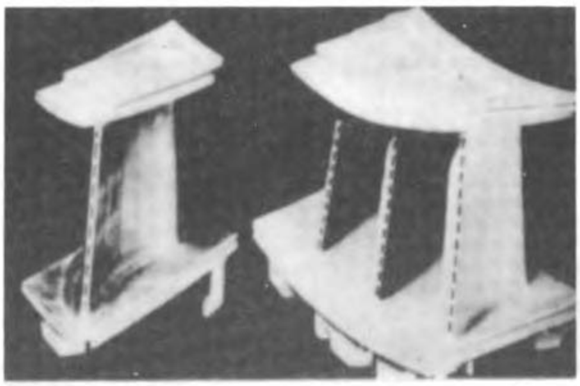


Figure 3-17. W501 turbine vane.

Although there are no centrifugal stresses on the vanes, the combination of gas bending loads and the thermal gradients caused by the vane cooling result in rather high localized steady-state operating stresses in stationary vanes. Thermal stresses from uneven heating and cooling of leading and trailing edges during starting and shutdown are also a factor and can cause cracking at these locations. The critical mechanical properties a vane alloy must possess are creep strength to resist distortion caused by gas loading and thermal stresses, low cycle fatigue strength to resist the cyclic thermal strains, and oxidation and corrosion resistance.

Material selection integrates advanced metallurgical concepts of alloy strengthening and material processing with the requirements of mechanical design and heat transfer. It is common to use the most advanced, highest strength alloy that also has the other attributes needed for vane use for the highly cooled first-stage vane. The design of the later stage vanes becomes a balance of two factors, alloy strength and amount of cooling. In some cases the choice is for high-strength alloys with little or no cooling, whereas in other cases, where moderate levels of cooling are used, lower strength alloys with greater castability are chosen.

Stationary vanes used for industrial gas turbines as currently built are single- or multiple-airfoil investment castings made from a cobalt-base alloy. In addition to the mechanical property requirements already discussed, the material must be easily cast into large (up to 150 lb), complex (internal cooling passages) configurations. A further requirement is weldability for ease of fabrication (cooling inserts are welded in place) and for repair of service-induced damage. The most popular vane alloys currently in use are Haynes Stellite Alloy No. 31 (X-40), its lower carbon derivative X-45, and a higher chromium cobalt alloy called FSX-414 (Table 3-2).

These alloys are precision-cast in air. The higher strength cobalt alloys, ECY-768 and MarM509, currently in use for some industrial turbine vanes, are vacuum-cast. Stress rupture properties for these cobalt-base turbine vane alloys are shown in Figure 3-18.

The cobalt-base alloys described are solid solution strengthened by the addition of the refractory metal elements tungsten and tantalum and by the formation of carbides of chromium and zirconium. Chromium is also important for the oxidation and corrosion resistance it imparts. The alloys are generally used in the as-cast condition or with an abbreviated heat treatment consisting of a solution treatment followed by an aging treatment to stabilize carbides. Occasionally fabrication processes such as brazing to secure the cooling inserts are incorporated in the heat treatment.

Table 3-2

NOMINAL COMPOSITIONS OF SELECTED ALLOYS USED FOR INDUSTRIAL GAS TURBINE VANES

Alloy	Co	Ni	Fe	Cr	W	Ta	Ti	C	B	Zr	Other
Cobalt base											
X-40 (Stellite 31)	Bal	10.0	1.5	25.0	7.5	--	--	0.5	--	--	--
X-45	Bal	10.5	2.0	25.0	7.0	--	--	0.25	0.01	--	--
FSX-414	Bal	10.5	2.0	29.5	7.0	--	--	0.25	0.012	--	--
Mar M 509	Bal	10.0	1.0	21.5	7.0	3.5	0.2	0.60	0.01	0.5	--
ECY-768	Bal	10.0	1.0	23.5	7.0	3.5	0.2	0.60	0.01	0.05	0.15 Al
Iron base											
Multimet N-155	20.0	20.0	Bal	21.0	2.5	--	--	0.15	--	--	3.0 Mo; 1.0 Cb
AISI Type 310	--	20.0	Bal	25.0	--	--	--	0.1	--	--	

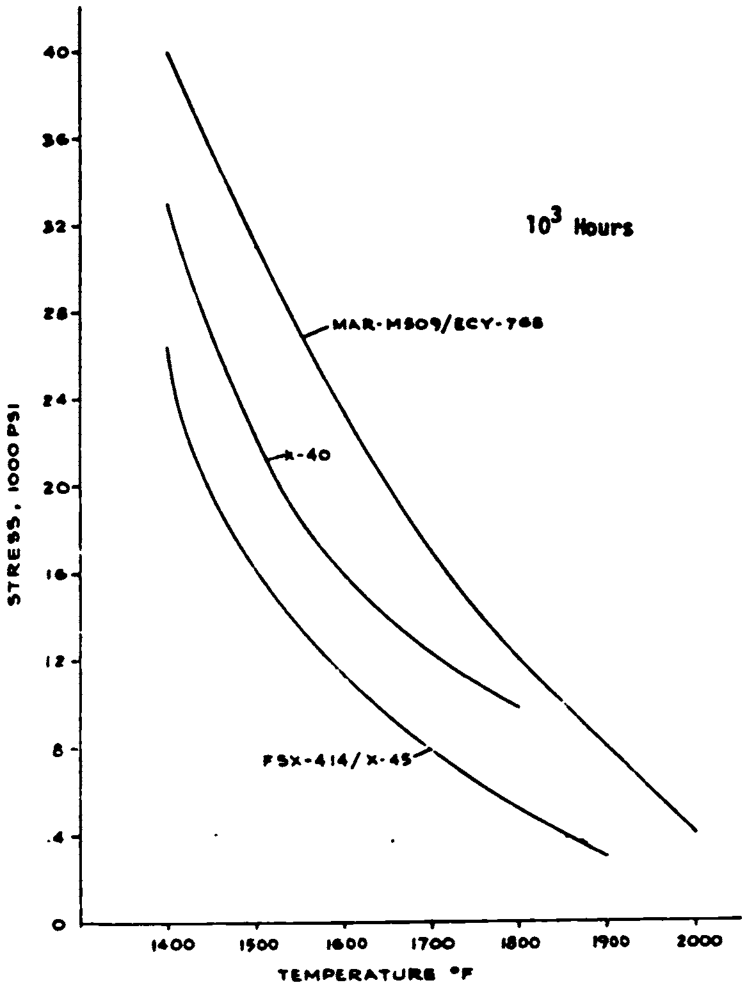


Figure 3-18. Stress-rupture properties of cobalt-base superalloys after 1000 hours (4).

Cast nickel-base alloys such as Udimet 500, IN-738, and IN-939, which are generally associated with turbine blades, have been used for some vanes. However, because it is difficult to produce blade-quality castings in large multivane segments, the nickel-base alloys have been used for single castings. A factor that could lead to greater use of nickel alloys for vanes would be a shortage in the availability of the strategic metal cobalt. Large three- and four-vane segments cast in N-155 have also been used for some cooler running last stages (temperatures in the 1000 to 1200°F range).

Repair of service-run vanes is an important consideration. This technique is routinely used to maximize the usefulness of these components. All of the alloys used are repairable by welding, although the difficulty increases with the strength of the alloy and the scope of repair becomes more limited. Occasionally a preweld heat treatment is necessary to return the material to a weldable metallurgical condition after extended high-temperature service. Gas tungsten arc welding is the usual choice, using a filler material of similar composition. Electron beam and plasma welding have also been employed. The scope of repair for vane segments is now being extended to include vacuum-brazing to fill minor thermal-fatigue and corrosion-related cracks.

It is anticipated that advancements in vane cooling technology, possibly in conjunction with minor improvements in vane alloy strength capability, will allow some uprating over the current state of the art. Beyond that, a second-generation industrial turbine will probably require a more significant change. The use of thermal barrier coatings to facilitate tolerance of higher temperatures and/or reduced cooling flow can have a pay-off approaching 200°F. Fabricated vane designs to introduce advanced, high-efficiency cooling schemes show promise. Materials such as oxide-dispersed products and single crystals also lend themselves to fabricated vanes, although considerable development effort will be necessary.

TURBINE BLADES

The turbine blades (buckets), shown in Figures 3-19 and 3-20, probably represent the most difficult materials application in the combustion turbine. Normal operating conditions include high temperatures and high centrifugal loadings, which are especially significant when one considers that the last-row blades in some of the larger turbines have vanes approaching 25 in. in length and the tips may be nearly 50 in. from the axis of rotation. In addition, there are low-level vibrations, thermal stresses from cooling, and thermal fatigue associated with starts and stops.

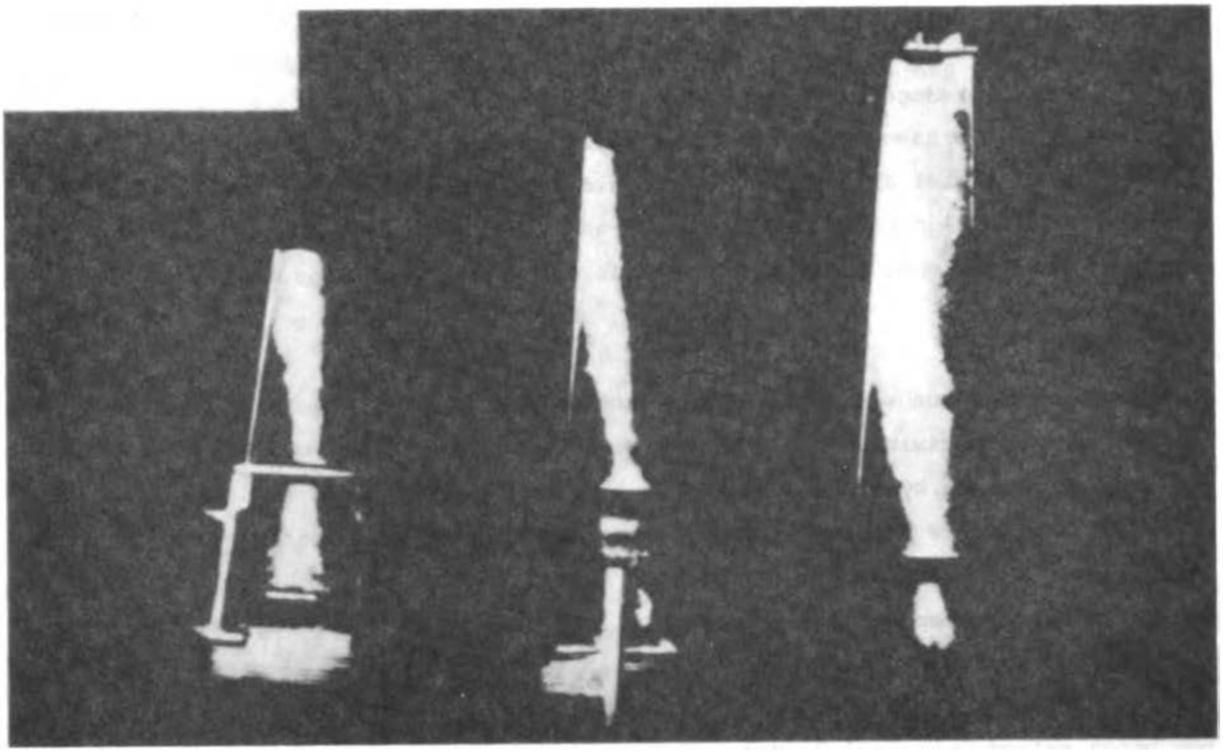


Figure 3-19. Gas turbine blades.

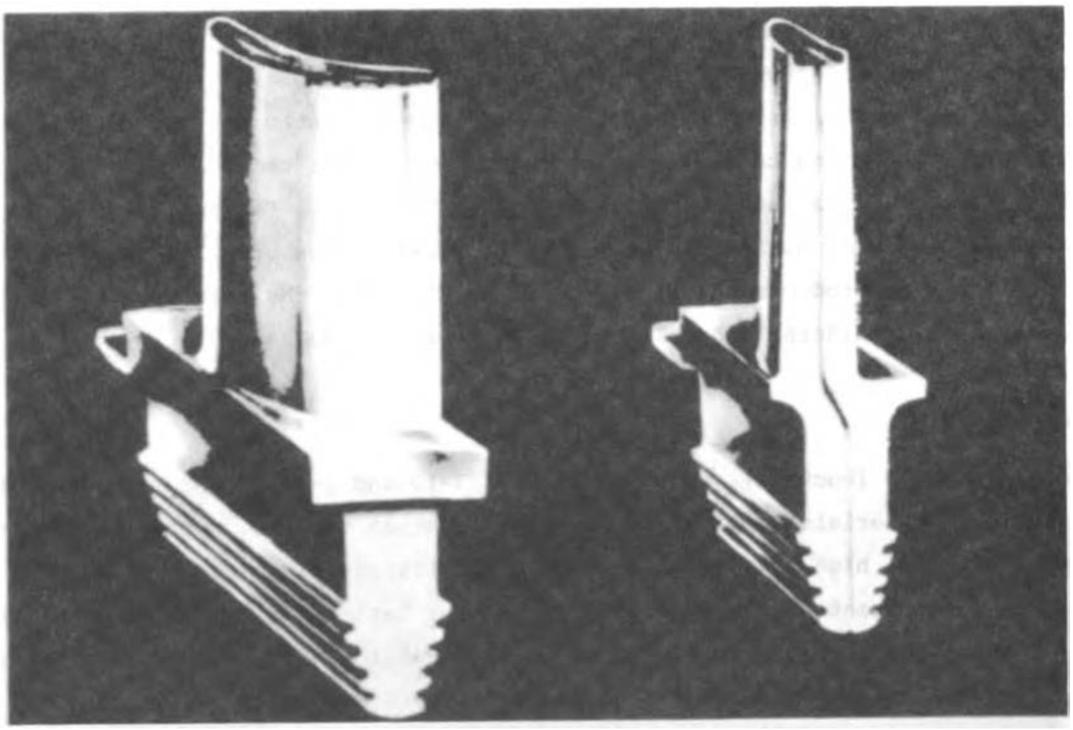


Figure 3-20. Air-cooled gas turbine blades.
(Source: Westinghouse Electric Corporation)

All of these concerns are valid for both aircraft and industrial combustion turbines. As a result, many of the same nickel-base alloys were used for both applications until it became apparent that there were unique problems associated with industrial turbine blading. In particular, these involve component size and resistance to hot corrosion from the action of condensed salts. Alloys such as B1900, MarM200, and Udimet 700, in part because of their relatively high aluminum and low chromium levels, possess excellent high-temperature strength and oxidation resistance. Consequently, they have been used extensively for aircraft turbine engine blading. However, the low chromium content also resulted in rather poor corrosion resistance at temperatures in the range of about 1200 to 1700°F, precisely the region where industrial turbine blades are expected to operate, often in a potentially corrosive combustion environment. To answer this need, a new family of higher chromium nickel-base alloys was developed and has been used successfully for industrial turbine blading. Among the more extensively used materials are the forging alloys Udimet 520, Udimet 720, and Inconel X-750 and the cast alloys Udimet 500, Inco 738, and GTD-111. Compositions of these alloys and some others used for aircraft turbine engines are listed in Table 3-3. Typical stress-rupture properties are given in Figure 3-21.

The processing methods used to fashion a turbine blade from a nickel-base superalloy have a large influence on the properties of the final product. This begins with the use of vacuum melting, vacuum arc or electroslag remelting, and vacuum casting of the product for conversion to forging bar or for master metal intended for vacuum investment casting. These processes permit tight compositional control, especially of the easily oxidized but critical strengthening ingredients, aluminum and titanium, and maintenance of very-low-gas contents. Precision forging and casting methods are utilized to produce the blades in near-finish sizes.

In the past, precision forging has been used to produce turbine blades in sizes ranging from 4-in. first-row blades to last-row blades approaching 30 in. This method is still used today for its ability to produce a high-quality, low-defect, dimensionally accurate, and stable part. Blades produced by precision forging in alloys like U-720 have creep strength nearly on a par with cast alloys and have somewhat superior high cycle fatigue resistance. Cooling passages are introduced into forged blades by drilling or electrochemical machining (ECM) radial holes that pass from the base of the blade to the tip.

With the use of complex ceramic cores, great flexibility in the design of cooling passages is possible using investment-cast blades. Developments in casting and

Table 3-3

NOMINAL COMPOSITIONS OF SELECTED NICKEL-BASE SUPERALLOYS USED FOR TURBINE BLADES

Alloy	Ni	Co	Cr	Mo	W	Ta	Al	Ti	C	B	Cb	Zr	Other	Form
Industrial turbine alloys														
Inconel X-750	Bal	--	15.5	--	--	--	0.7	2.5	0.04	--	1.0	--	7.0 Fe	Wrought
U-520	Bal	12.0	19.0	6.0	1.0	--	2.0	3.0	0.05	0.005	--	--	--	Wrought
U-710	Bal	15.0	18.0	3.0	1.5	--	2.5	5.0	0.07	0.02	--	--	--	Wrought
U-720	Bal	15.0	18.0	3.0	1.25	--	2.5	5.0	0.035	0.035	--	0.035	--	Wrought
U-500	Bal	18.0	19.0	4.0	--	--	3.0	3.0	0.08	0.005	--	--	--	Cast (Wrought)
IN-738	Bal	8.5	16.0	1.75	2.6	1.75	3.4	3.4	0.10	0.01	0.9	0.05	--	Cast
IN-939	Bal	19.0	22.5	--	2.0	1.4	1.9	3.7	0.15	0.01	1.0	0.1	--	Cast
GTD-111	Bal	9.5	14.0	1.5	3.8	2.8	3.0	4.9	0.10	0.01	--	0.03	--	Cast
Aircraft turbine alloys														
IN-713C	Bal	--	12.5	4.2	--	--	6.1	0.8	0.12	0.012	2.0	0.1	--	Cast
U-700	Bal	18.5	15.0	5.0	--	--	4.4	3.5	0.07	0.025	--	--	--	Wrought (Cast)
B-1900	Bal	10.0	8.0	6.0	--	4.3	6.0	1.0	0.1	0.015	--	0.08	--	Cast
IN-100	Bal	15.0	10.0	3.0	--	--	5.5	4.7	0.18	0.014	--	0.06	1.0 V	Cast
Mar M 200	Bal	10.0	9.0	--	12.5	--	5.0	2.0	0.15	0.015	1.8	0.05	--	Cast
R80	Bal	9.5	14.0	4.0	4.0	--	3.0	5.0	0.17	0.015	--	0.03	--	Cast

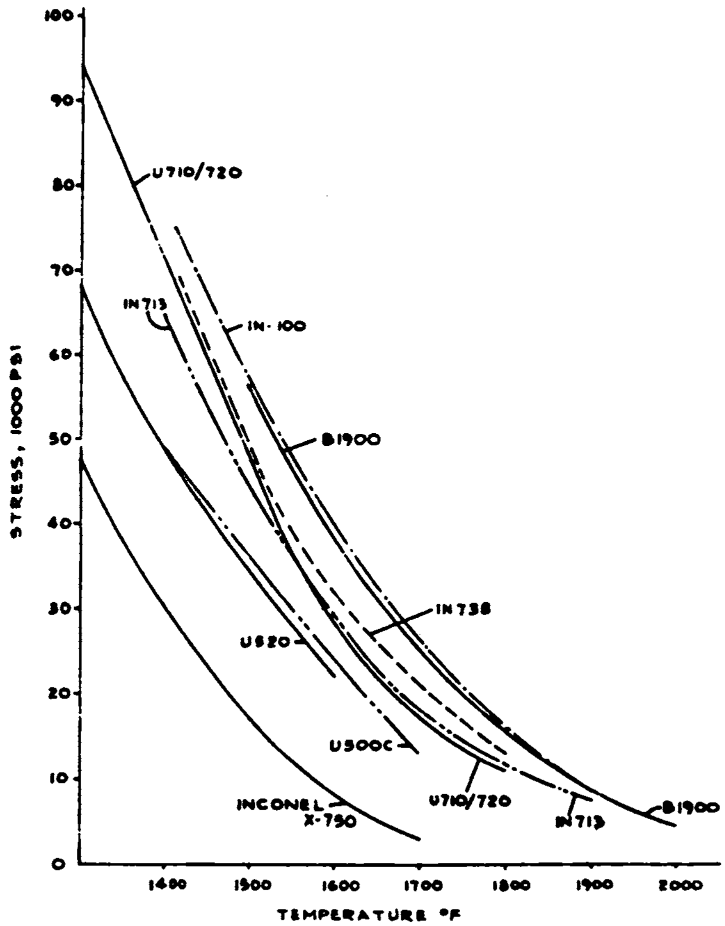


Figure 3-21. Stress-rupture properties of nickel-base superalloys after 1000 hours (4).

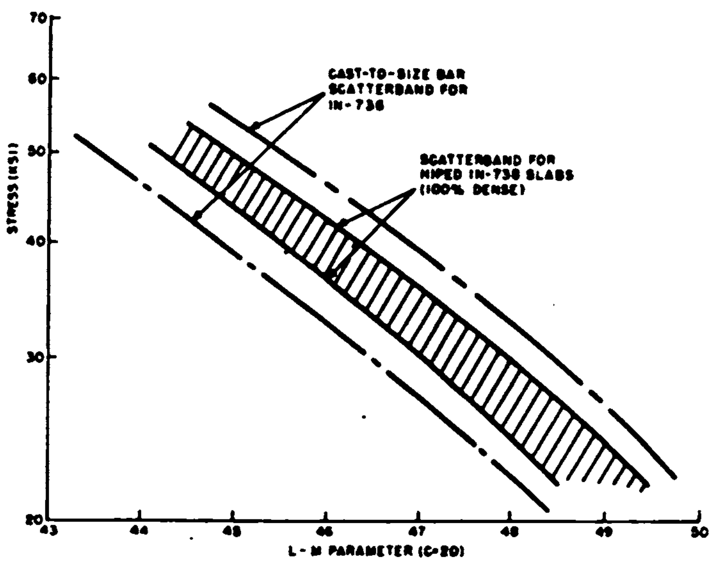


Figure 3-22. Comparison of stress-rupture properties for as-cast and hot isostatic pressed IN-738, all heat-treated (4).

solidification control over the last 10 to 15 years have made it possible to produce cast blades with dimensional accuracy approaching the forged blades. Cast alloys can be more highly alloyed than forging (wrought) alloys and consequently are capable of greater high-temperature creep strength. The cast metallurgical structure also tends to produce slightly greater creep life than the same alloy composition in wrought form (U-500, for example, is available in both forms). Unfortunately, the more highly alloyed cast materials such as IN-738 can contain casting defects (e.g., porosity) that cause a variability in properties in excess of the spread normally anticipated for wrought materials. Current practice often includes hot isostatic pressing (i.e., the simultaneous application of isostatic pressure and temperature) to close porosity and reduce the property variability. This procedure is effective in improving the fatigue properties of some cast blades and reducing the stress-rupture scatter, as shown in Figure 3-22. The mechanical properties of cast alloys are also influenced by the cross section of the part, since this has a strong effect on cooling rate and therefore microstructure. Care must be taken to design parts using cast alloy properties derived from specimens that are representative of the dimensions of the blade being considered.

An important point to keep in mind is that all precipitation-hardened nickel-base blade alloys are, by nature, slightly unstable at operating temperatures. Overaging (coarsening) of the gamma prime precipitation is a thermodynamically favored process that causes a reduction in creep strength.

These events are accelerated by stress. Alloy compositions are balanced to retard this process as much as possible so that long service lives can be obtained without serious loss of strength. In some alloys, reapplication of the full heat treatment has been shown to return the service-aged and weakened microstructure approximately to its initial condition and strength. When second-stage creep has progressed to the point of void formation at internal grain boundaries, but without appreciable macrodeformation, hot isostatic pressing (HIP) has been used in conjunction with heat treatment to restore properties (4). This use of HIP is analogous to the closing of internal voids in as-cast blades.

Alloy composition is also adjusted to minimize the possibility of brittle-phase formation during service exposure in certain temperature ranges. The embrittling sigma phase has been observed in some nickel-base alloys after service in the 1550 to 1700°F range (5). Research has shown that several elements--notably chromium, tungsten, and molybdenum--can promote sigma formation, as can the application of stress. It is obvious that indiscriminately adding chromium to maximize hot

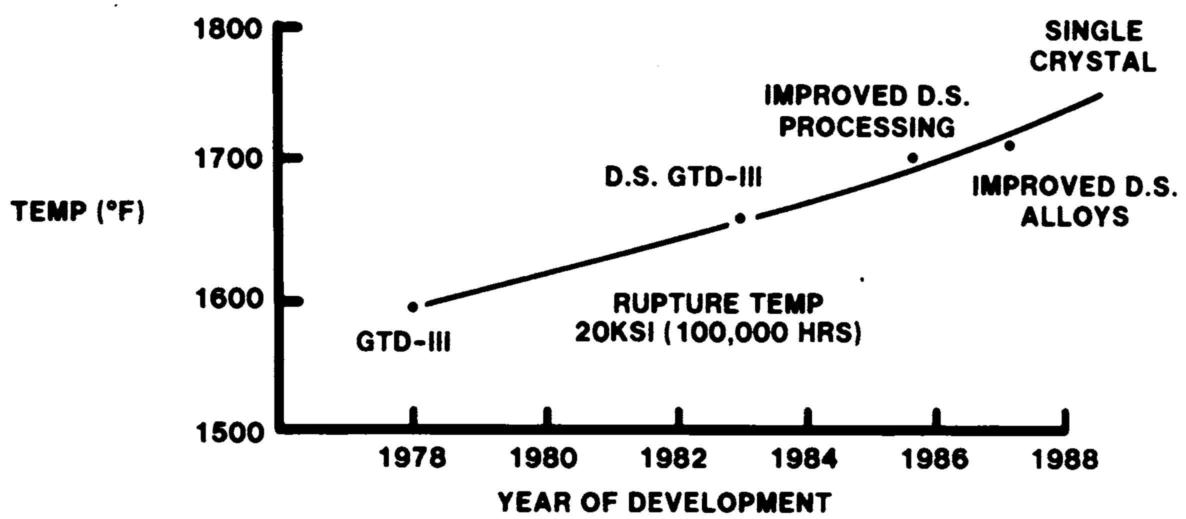


Figure 3-23. Future blade materials and processes.

corrosion resistance could result in an alloy that is rapidly embrittled. Careful balancing with elements that retard sigma formation was found to be the answer. Today, turbine blade alloys are made to a controlled composition (e.g., Phacomp control), and the analysis of individual heats is compared to a standard that is not sigma-prone.

The most highly alloyed, highest strength nickel-base alloys are used for the hotter first- and second-stage blades. Later-stage blades, while cooler, are considerably larger and present different problems in alloy selection. Along with rupture and creep strength, these blades must resist high tensile stresses in the blade root caused by centrifugal loading and must also have good resistance to high cycle fatigue. Forged blades made from Udimet 520 and Inconel X-750 or cast blades of Udimet 500 are used for this application.

It is believed that the most feasible path for improving early-stage blade creep strength is through the extension of the directional solidification process, currently in use for aircraft blades, to the very much larger land-based gas turbine blades. With the application of this technology a development path as shown in Figure 3-23 is considered realistic. Actually, the feasiblity of producing large-size blades has already been demonstrated (Figure 3-24), and property improvements comparable to those achieved in aircraft blades have been observed. Before undertaking production application of this process, development work is needed to optimize blade designs (e.g., reduced section sizes) for economic reasons and to perfect casting equipment, processing, and alloys for the larger size blades needed.

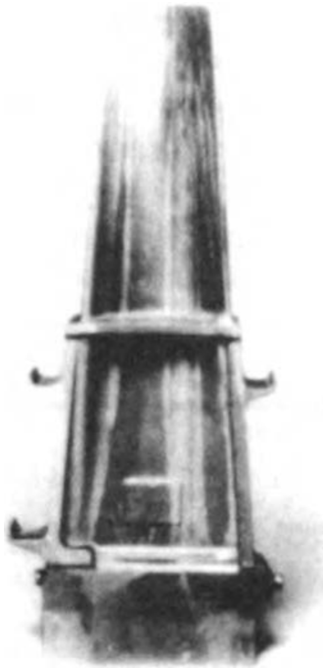


Figure 3-24. Directionally solidified MS7000E first-stage blade.

Beyond the application of directional solidification, single-crystal alloys are attractive not only because of further property enhancement but also because of the alloy design flexibility offered. Since consideration of grain boundary strengthening is not required, it is possible to achieve improved corrosion resistance in parallel with creep strengthening. There is a question about the capability of producing large single-crystal blades, and therefore feasibility studies should be initiated as soon as possible to answer this question. With the application of directional solidification and possibly single-crystal technologies, improvements of 200°F above current operating temperatures in blade capability are believed attainable. With the application of advanced blade cooling designs as used in aircraft engines and the possible application of thermal barrier coatings to blades, the goals stated in this study should be attainable.

These advances will also be accompanied by increased mass flow, with a corresponding increase in exit annulus area requiring larger last-stage blades. Current alloys are difficult to cast, particularly in larger sizes. For this reason, work is needed for the development of last-stage blade alloys having increased castability or for the application of stronger forged last-stage blades.

As with turbine vane segments, repair of service-induced defects and restoration of properties through reheat treatment is being used to extend the usefulness of turbine blading. The creep, impact, and other properties of the blade alloys are degraded by long-term service. These properties can be restored for many of the alloys to near the original levels by the application of a multiple-stage heat treatment. Because of concern over the occurrence of weld defects, undetected heat-affected zone strain age cracking, and lack of filler materials matching the properties of the base material, weld repair of blades is currently limited to low-stress areas (e.g., near the tip). Consequently, while blade reheat treatment is widely accepted for most alloys, weld repair of blades is more sparingly employed, especially compared to vane segment repair. Further development work is needed in the area of blade repair processes, including design substantiation for processes to be used.

COATINGS

Coatings are applied to blades and in some cases to vanes to resist surface degradation from hot corrosion attack. Application of coatings in land-based turbines occurred in the mid-1970s following the development of higher firing-temperature turbines for increased output and efficiency. In the development of superalloys for these applications, it has become increasingly difficult to design alloys having both the desired increased high-temperature strength, stability, and desired corrosion resistance. Emphasis has therefore been placed on the development of reliable, long-life, corrosion-resistant coatings. Early coatings tried were those already in use in the aircraft industry for oxidation resistance. However, these coatings were soon found inadequate for industrial use because the air and fuels encountered frequently contain corrosive contaminants.

Hot corrosion is a rapid form of attack resulting from the condensation of alkali sulfates formed during combustion onto blade or vane surfaces, causing a dissolution of the protective oxide layer at temperatures between 1500 and 1700°F. With the breakdown of the protective oxide, sulfur is able to diffuse into the alloy surface layer where chromium sulfides can form. The by-products of this reaction eventually prevent the rapid formation of an adherent protective oxide and result in very rapid attack. This mechanism of hot corrosion is referred to as hot sulfidation or Type I hot corrosion.

A second type of attack, referred to as Type II, or low-temperature hot corrosion, occurs in the 1100 to 1400°F range. It is caused by the formation of low-melting eutectic compositions of alkali metal sulfates and base alloy metal sulfates such as

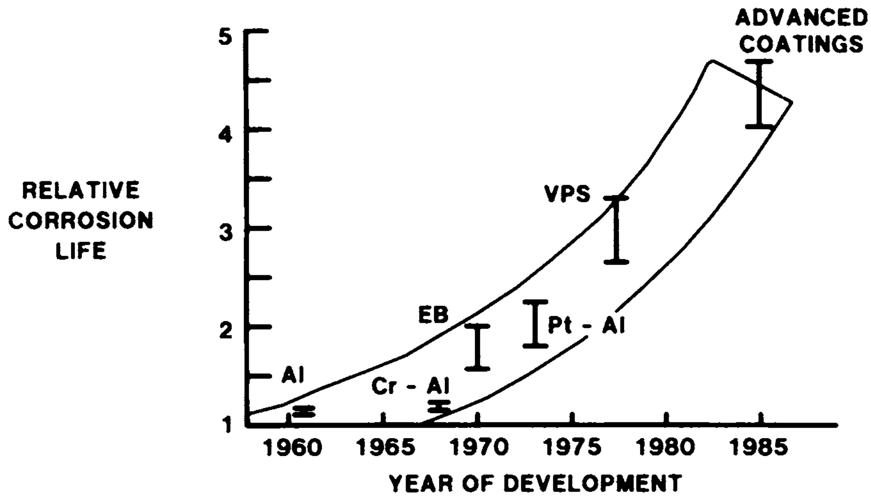


Figure 3-25. Coating chronology (3).

nickel and cobalt that result in dissolution of the protective surface oxide layer. At the temperatures of Type II attack, internal sulfidation is not normally observed. This type of corrosion is analogous to that observed in coal-fired boilers and referred to as "fireside corrosion."

Defenses against either form of hot corrosion are, first, to reduce the contaminants in the fuel and air; second, to use base materials that are as corrosion-resistant as possible; and third, to apply corrosion-resistant coatings.

The development of early corrosion-resistant coatings was aimed at combatting Type I corrosion. The chronology of this development is shown in Figure 3-25. Diffusion coatings such as platinum aluminides were among the first to be used and have been very successful for reducing hot corrosion damage during the past 10 years. However, these coatings are thin (0.002 to 0.004 in.) because of the diffusion method of application. Further, the chemistry of these coatings is not readily modified for further improvement in corrosion resistance. For these reasons, increased attention was given to the development of overlay coatings. There are various ways to apply overlay coatings, among which are the electron beam (EB) vapor deposition and vacuum plasma spray (VPS) processes. Figure 3-26 compares the structures obtained using conventional atmospheric plasma-sprayed and vacuum plasma-sprayed coating. The dramatically reduced oxides together with the high density and the high-quality bond line of the VPS coating are very apparent. Both the EB and VPS processes are now successfully in production, and extensive coating development activity is being carried on for improved compositions.

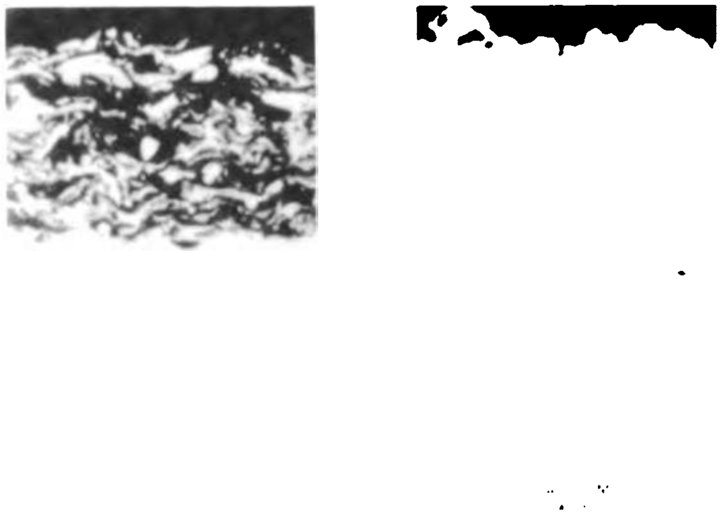


Figure 3-26. Comparative microstructures.

Currently, because of the increased thickness readily attainable with overlay processes, coatings having twice the life of the platinum aluminide diffusion coatings are now being applied.

Coatings developed for protection from Type I corrosion have been found inadequate for protection against Type II attack. Development efforts aimed at improved coating compositions have received increased attention during the past 5 years. Increasing the chromium content of the coating has been found very effective in protecting against Type II attack.

As turbine temperatures are increased further, even with the use of projected clean coal-derived gas fuels, it is expected that blade coatings will continue to be used and even extended to downstream stages, for the following reasons: First, early-stage metal temperatures will be increased, requiring increased oxidation resistance; second, it has been found that nickel alloy environmental embrittlement in service is significantly reduced with the presence of a coating; third, it is believed that continued corrosion protection will be required because of upsets in the coal gasification and clean-up process and because of contaminants introduced with inlet air.

Another class of coatings now being applied to combustor components consists of the thermal barrier coatings (TBCs). This technology, transferred from the aircraft engine industry, provides the benefit of reduced metal temperatures of air-cooled components; perhaps even more important, TBCs smooth out hot streaks or spots in high-temperature components, thus reducing thermal fatigue stresses. A large benefit exists in the application of TBC technology to vanes and blades. To achieve this it is necessary that reliably adherent and smooth coatings be developed along with an oxidation- and corrosion-resistant capability.

ROTORS

Compressor rotors have been made by shrinking disks onto forged shafts, bolting disks together (Figure 3-27), or by welding disks together at the outer diameter to form a monolithic construction. Disk alloys currently in use are shown in Table 3-4. Because of the relatively low temperature of the compressor (up to approximately 750°F), it is possible to use steel for the disks. A major consideration for the material selection and processing is protection against brittle fracture. Because of the large section sizes (up to 14 in.), a quenched and tempered vacuum-degassed NiCrMoV steel is commonly used on stages where the disk service temperature is below the onset of temper embrittlement (approximately 600°F). Beyond this temperature, material selection and processing follow practices used for turbine disks (i.e., CrMoV, 12 Cr, hot spinning).

It is general practice to use vacuum-degassed steel for the manufacture of NiCrMoV

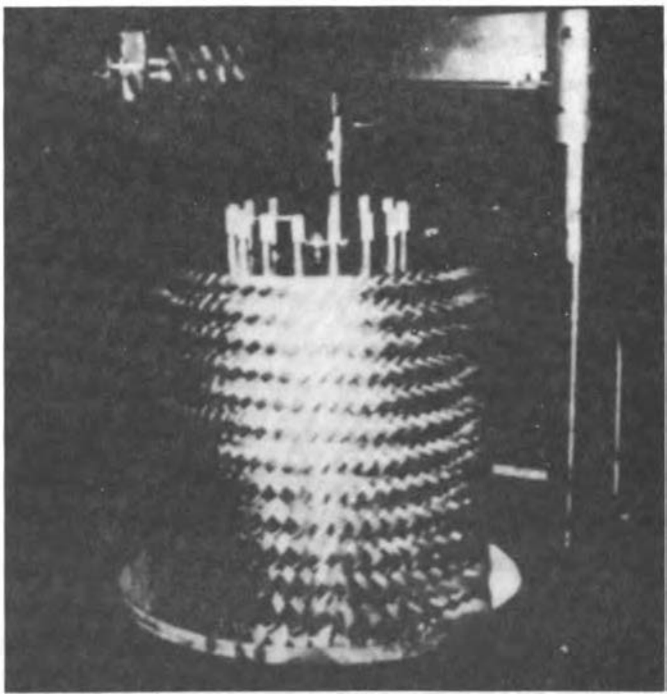


Figure 3-27. Stacking compressor wheels or disks (1).

Table 3-4

USE OF IRON-BASE ALLOYS IN INDUSTRIAL GAS TURBINES

Component	Alloy	Nominal Composition
Casing	Carbon steel plate (e.g., A-515 Gr. 70)	Fe; 0.3 C; 0.75 Mn; 0.25 Si
Inlet, compressor,	Carbon steel casting (e.g., A-356 Gr. 1)	Fe; 0.3 C; 0.8 Mn; 0.5 Si
and combustor	Cast iron (e.g., A-278)	Fe; 3.8 total C max.
Turbine	Low-alloy steel plate (e.g., A-387 Gr. 22)	Fe; 2.25 Cr; 1.0 Mo; 0.15 C
	Low-alloy steel castings (e.g., A-217 Gr. WC9)	Fe; 2.25 Cr; 1.0 Mo; 0.15 C
	Nodular iron (e.g., A-395)	Fe; 3.0 total C min.; 2.5 Si
Compressor blades	AISI Type 403	Fe; 12 Cr; 0.12 C
Compressor vanes	AISI Type 403	Fe; 12 Cr; 0.12 C
Discs	Low-alloy steel (e.g., AISI-4140)	Fe; 1.0 Cr; 0.2 Mo; 0.4 C
Compressor	Low-alloy steel (e.g., AISI-4340)	Fe; 2.0 Ni; 0.75 Cr; 0.25 Mo; 0.4 C
	Low-alloy steel (e.g., A-471 Gr. 10)	Fe; 1.2 Cr; 1.15 Mo; 0.25 V; 0.3 C
	Super 12 chrome steel (e.g., AISI Type 422)	Fe; 0.5 Ni; 12 Cr; 1.1 Mo; 0.3 V; 1.1 W; 0.75 Mn; 0.5 Si; 0.25 C
	Super 12 chrome steel (e.g., FV535)	Fe; 0.5 Ni; 6.0 Co; 11.0 Cr; 0.75 Mo; 0.25 V; 0.4 Cb; 0.9 Mn; 0.5 Si; 0.09 C
Turbine	Low-alloy steel (similar to A-471)	Fe; 3.5 Ni; 1.75 Cr; 0.5 Mo; 0.1 V; 0.35 C;
	Low-alloy steel (similar to A-471)	Fe; 1.2 Cr; 1.15 Mo; 0.25 V; 0.3 C
	Iron-base superalloy (e.g., Discalloy)	Fe; 26 Ni; 13.5 Cr; 2.75 Mo; 1.75 Ti; 0.1 Al; 0.04 C; 0.9 Mn
	Iron-base superalloy (e.g., A-286)	Fe; 26 Ni; 15 Cr; 1.25 Mo; 0.3 V; 2.15 Ti; 0.2 Al; 0.05 C; 1.4 Mn
Turbine vanes	Super 12 chrome steel (M-152)	Fe; 2.5 Ni; 12 Cr; 1.7 Mo; 0.3 V; 0.12 C
	Multimet (N-155)	Fe; 20 Ni; 20 Co; 21 Cr; 3 Mo; 2.5 W; 1 Cb; 0.15 C; 1.5 Mn
	AISI Type 310	Fe; 20 Ni; 25 Cr; 0.1 C

steel to avoid hydrogen flaking. Magnetic particle and ultrasonic inspection are used to check for critical size defects.

Turbine disks are generally of a bolted construction. The use of cooling and long-shank blades resulting in lower bore and disk dovetail temperatures has permitted the continued use of steel turbine disks. A common alloy used is quenched and tempered 1 Cr 1 Mo 1/4 V steel, which is the same composition used for high-pressure steam turbine rotors. The use of this bainitic steel is based on its excellent creep strength. However, its hardenability is low, resulting in low toughness in heavy sections. To overcome this, one practice used is to prespin wheels above the brittle-to-ductile transition temperature to provide residual compressive stresses in the wheel bore and thus reduce the bore tangential tensile stresses during service. Alternatively, NiCrMoV steel, because of its high toughness, is sometimes used for turbine disks, provided temperatures are kept sufficiently low to avoid temper embrittlement and to accommodate the lower creep strength of this class of steel. CrMoV and NiCrMoV steels used for turbine disks are either vacuum-arc remelted or electroslag remelted to reduce deleterious effects of segregation, primarily of sulfur. Low-sulfur (0.005 percent max.) vacuum-degassed steel (ladle or stream) is also used, resulting in bore ductility competitive to that achieved using remelting processes.

Turbine disks are also made of 12 Cr steels. Not only do these steels have higher creep strength compared to CrMoV steel, but they are fully hardenable, and compositions such as M-152 (12 CrNiMoV) are available having excellent toughness throughout thick sections. Consumable remelting processes are used to produce the 12 Cr disk steels to reduce segregation effects.

Austenitic disk materials such as Discalloy and A-286 are used in some applications. These materials are vacuum-arc remelted, and because of size limitations that have existed in their manufacture and because of their higher cost they are currently not used extensively. However, with future increases in turbine temperature, increased use of austenitic turbine disks will probably be necessary. Large development disks have been made using materials such as IN-706 (7), IN-718, and IN-901. Both vacuum-arc remelting and electroslag remelting have been used, the latter being preferred because of less segregation and therefore larger ingot size capability.

Representative compressor-turbine rotor assemblies are shown in Figures 3-28 and 3-29.

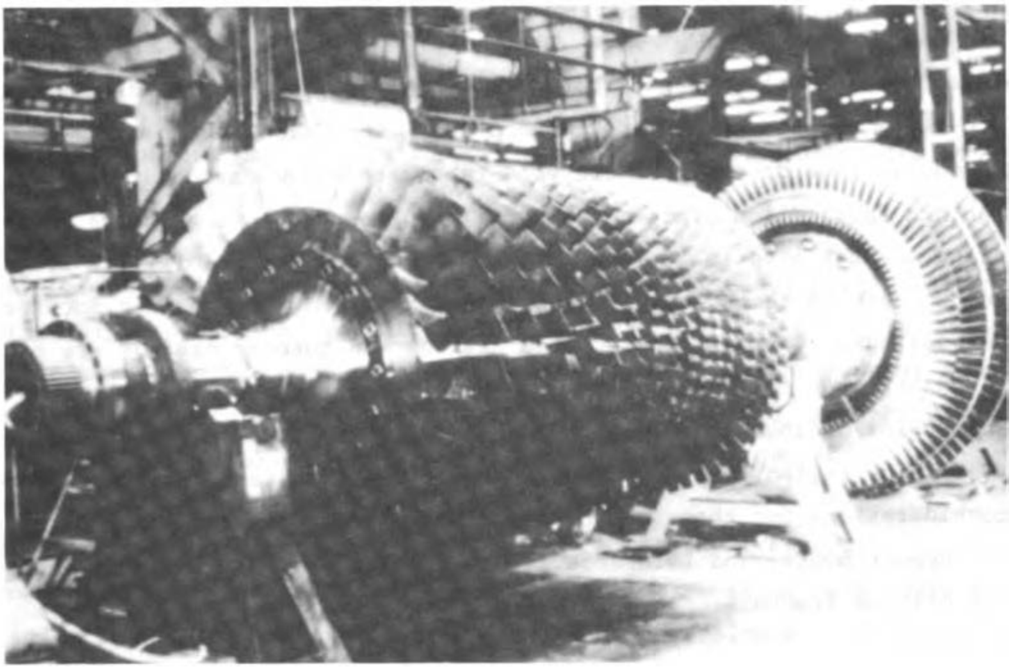


Figure 3-28. Turbine compressor bolted rotor assembly.

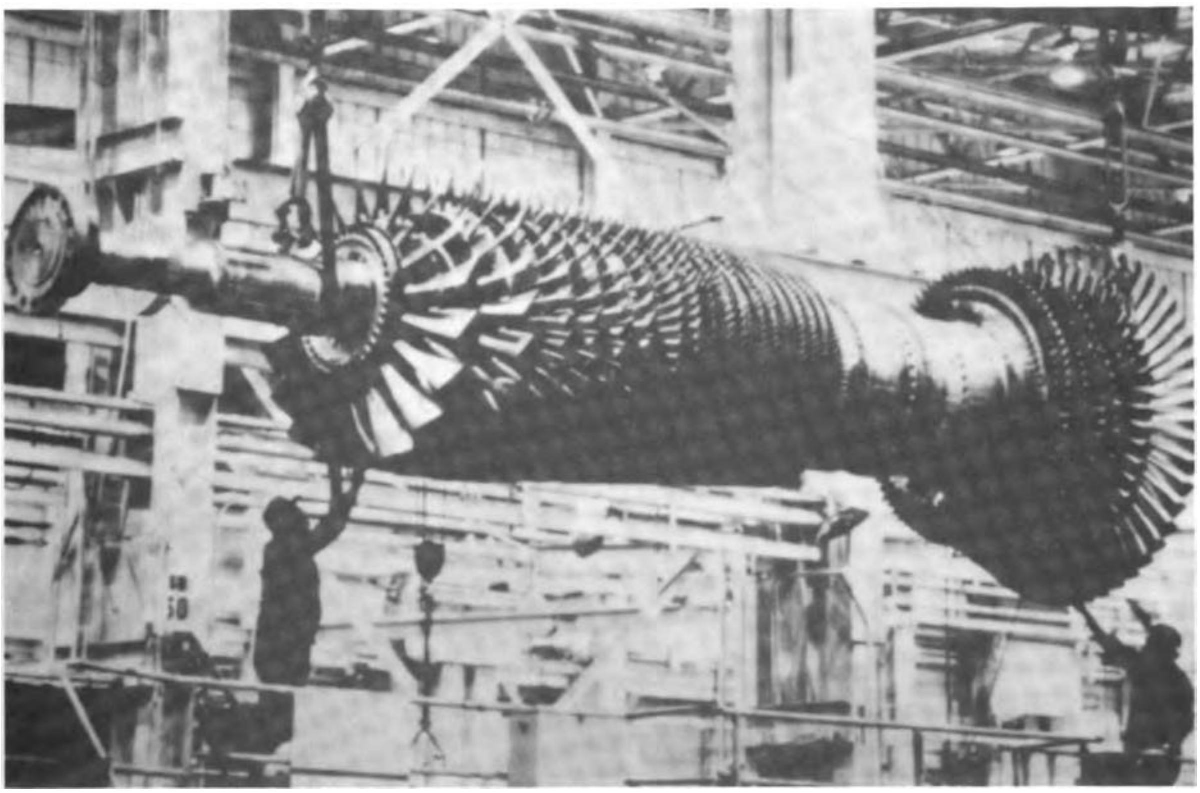


Figure 3-29. Complete rotor ready for installation.

CASINGS

The casings of a land-based gas turbine are a horizontally split and bolted design similar to steam turbine practice (Figure 3-30). They are cast in grey iron or ferritic nodular iron or steel or are weld-fabricated from carbon or CrMo steel. The design and fabrication practices generally follow the ASME pressure vessel code. An advantage of cast iron is its low cost and excellent producibility. Grey iron is restricted to service temperatures below about 450°F and thus is used for forward compressor casings. Compressor discharge casings and turbine casings are made from ferritic nodular iron for service up to about 650°F or from welded carbon or CrMo steel. Combustor casings and wrappers are also fabricated from carbon or CrMo steel. Very high reliability has been experienced with these practices. Major design considerations for these casings are thermal fatigue crackings, particularly at nozzle support hooks, and tolerance for defects to prevent low cycle fatigue growth and brittle fracture.

It is believed that these practices will continue to be used with increasing turbine

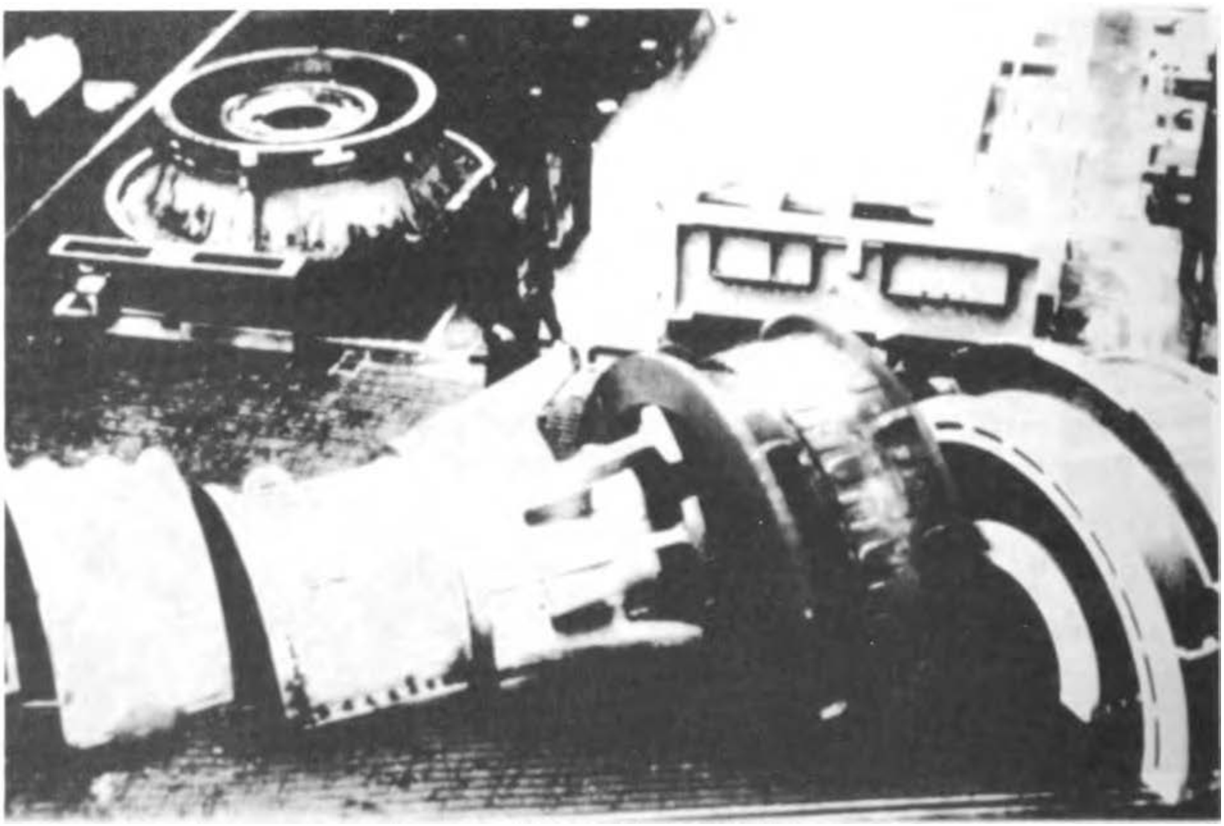


Figure 3-30. Gas turbine cast casings.

firing temperatures. However, increased air cooling and greater use of alloy steel rather than cast iron in the hot sections will be necessary.

INTERNATIONAL GAS TURBINE RESEARCH AND DEVELOPMENT

Research and development programs in support of advanced gas turbines are being performed both in Europe and Japan.

In 1971, a European Collaborative Program on Materials for Gas Turbines was initiated (8,9). This program, called COST-50, is supported by the Commission of European Communities. Nine European countries (Austria, Belgium, the Federal Republic of Germany, France, Italy, the Netherlands, Sweden, Switzerland, and the United Kingdom) and the Joint Research Center of the Community are involved.

This program is aimed at advancing the materials technology primarily for the gas path components. Major tasks include corrosion studies, coating development, creep, fatigue, and structural stability studies, and process development including welding, forging, casting, and powder metallurgy. There are no known plans to develop an advanced gas turbine design under the program. This will presumably be done by the European turbine manufacturer.

In Japan, a program known as the Moonlight Project was initiated in late 1978 under the sponsorship of the Agency of Industrial Science and Technology, with a major role played by the Engineering Association for Advanced Gas Turbines. The program consists of the following four subprograms:

- Research and development of high-temperature heat-resistant materials (heat-resistant alloys and ceramics, scheduled through 1984).
- Research and development of component technology (compressors, combustors, turbines, control systems, and heat exchangers, scheduled through mid-1986).
- Advanced gas turbine test operation (pilot plant scheduled through 1984 and a prototype plant scheduled from 1983 through 1987).
- Evaluation of total energy systems (scheduled through 1987).

The pilot plant consists of a reheat gas turbine utilizing two phases of combustion and a water-spraying evaporative intercooler for compressor air. Two compressor and turbine sections are employed using a two-shaft system. The low-pressure compressor contains 10 variable stages, and the high-pressure compressor contains 16 nonvariable stages. The total pressure ratio is 55:1, with an airflow of 220 kg/sec. The firing temperatures of the high-pressure and low-pressure sections are 1300°C and

1200°C respectively. The specified output is approximately 100 MW, and the combined plant efficiency is 50 percent (LHV basis) and 45 percent (HHV basis) (10).

Highlights of advanced materials being evaluated are directionally solidified turbine blades, silicon nitride and silicon carbide tiled combustors, slip-cast reaction-bonded silicon nitride transition pieces, use of ODS diffusion-bonded MA754 alloy, and advanced cooled blades and vanes (film and impingement cooling) (11,12,13,14).

REFERENCES

1. D. E. Brandt. Gas Turbine Philosophy. General Electric 1984 Gas Turbine Reference, GER-3434.

2. Advanced Cooling Full Scale Engine Demonstration Program. EPRI Report Under Contract RP1319-1, Aug. 1981.

3. F. D. Lordi, A. D. Foster, and P. W. Schilke. "Advanced Materials and Coatings", General Electric 1984 Gas Turbine Reference Library, GER-3421.

4. C. T. Sims and W. C. Hagel, eds. The Superalloys. New York: John Wiley and Sons, 1972.

5. J. P. Dennison, I. C. Elliott, and B. Wilshire. "An Assessment of Hot Isostatic Pressing and Reheat Treatment for the Regeneration of Creep Properties of Superalloys," pp. 671-677 in Superalloys 1980, J. K. Tien et al., eds. New York: ASM, 1980.

6. G. E. Washielewski and N. R. Lindblad. "Elimination of Casting Defects Using HIP." Superalloys: Processing, Proceedings of the Second International Conference.

7. A. M. Johnson and K. E. Fritz. "The Properties and Microstructure of a Large Forged Superalloy Turbine Wheel." Superalloys: Metallurgy and Manufacture, Proceeding of the Third International Conference.

8. D. Coutsouradis, P. Felix, H. Fischmeister, L. Habraken, Y. Lindblom, and M. Speidel, eds. High Temperature Alloys for Gas Turbines. Applied Science Publishers, 1978.

9. R. Brimetand, D. Coutsouradis, T. Gibbons, Y. Lindblom, D. Medoncroft, and R. Stickler, eds. High Temperature Alloys for Gas Turbines. Reidel, 1982.

10. Advanced Reheat Gas Turbine Pilot Plant AGTJ-100A. Toyko: Engineering Research Association for Advanced Gas Turbines.

11. S. Aoki, K. Techima, M. Arai, and I. Fukue. "Aerodynamic and Cooling of a Full Scale High Temperature Turbine." ASME IGTC-116, 1983.

12. K. Mori, J. Kitajima, T. Kimura,and Y. Enzaki. "Research and Development on Combustors for Advanced Reheat Gas Turbine." ASME IGTC-21, 1983.

13. H. Yasui. "Performance and Characteristics of Twin Shaft Reheat Cycle Gas Turbine." ASME IGTC-40, 1983.

14. K. Takeya and Y. Oteki. "Technical Problems on Advanced Reheat Gas Turbines Under the Moonlight Project." ASME IGTC-117, 1983.

Section 4

MATERIALS AND PROCESSES FOR ADVANCED LARGE LAND-BASED GAS TURBINES

This section reviews the current status of materials and processes for advanced large land-based gas turbines and indicates what further work is needed to adapt these materials and processes to the turbines to be developed over the next 15 to 20 years. The topics discussed were selected as representing the primary considerations in designing and developing future gas turbines. They do not specifically include some topics that are indeed important (e.g., bearings, pumps, filters, burner nozzles, controls, housings, insulation). It is believed, however, that the most critical items are included.

DIRECTIONALLY SOLIDIFIED AND SINGLE CRYSTAL SUPERALLOYS

A principal consideration in the selection of materials for high-pressure (HP) turbine blades in most aircraft gas turbines is high creep-rupture strength. In the more advanced, high-performance turbines, excellent creep-rupture properties have been achieved through the use of directionally solidified (DS) (columnar-grained) and single crystal (SC) nickel-base superalloy HP blades (1,2). In addition, as will be discussed below, DS and SC blades have further benefits, including high thermal fatigue resistance and, in the case of single crystals, enhanced alloying flexibility for oxidation and hot corrosion resistance. As performance requirements increase, it would be desirable to take advantage of these properties for large land-based gas turbines. However, because components in such turbines tend to be on the order of 3 to 4 times as large as their aircraft counterparts, the castability of components in this size range requires particular attention. Furthermore, if extraordinary measures are needed in order to cast satisfactory components, these might affect the economic viability of utilizing this class of materials.

High-pressure turbine blades in early versions of aircraft gas turbines were fabricated from wrought nickel-base superalloys. With improvements in vacuum melting and casting practice, higher strength cast nickel-base superalloys were introduced into service. The creep-rupture lifetime of both the wrought and cast blades was limited, in part, by cracking of grain boundaries normal to the direction of centrifugal loading. To minimize or completely preclude intergranular creep cracking, the concept of directional solidification was introduced by VerSnyder (3). With this approach, the component consists of an array of grains, all parallel to the principal direction of loading. The grain boundaries are also aligned with the loading direction. Significant improvements in creep strength (to 1 or 2 percent strain) and rupture lifetime were achieved with this revolutionary development.

For several reasons it was decided to extend this development one step further by completely eliminating the grain boundaries--i.e., growing a single crystal. The reasons included the following:

1. The directional solidification process resulted in large MC-type carbides in some alloys, and these carbides were often precracked and initiated matrix fatigue cracks under cyclic loading conditions (4,5).

2. Some grain boundaries were not perfectly aligned with the solidification direction, and creep cracks were initiated at these boundaries where they intersected a free surface.

3. The removal of grain boundary strengthening elements, including C, B, Zr, and Hf, might improve the fatigue properties by elimination of MC carbides and could increase the incipient melting temperature and therefore the creep resistance, because these elements are melting-point depressants.

4. In the absence of grain boundaries, more flexibility in alloying might be achieved that would result in an optimum balance of creep-rupture strength and oxidation and hot corrosion resistance.

In fact, single crystals were developed that have superior properties compared to the available DS alloys. The comparative creep strength, thermal fatigue resistance, and oxidation resistance for conventionally cast B1900 + Hf, DS Mar-M200 + Hf, and SC PWA 1480 are shown in Figure 4-1. Single crystal turbine blades were put into commercial service in 1982 in one manufacturer's engines on the Boeing 767 aircraft.

Figure 4-1 also shows that state-of-the-art DS and SC superalloys generally have far superior thermal fatigue resistance compared to a state-of-the-art conventionally cast (CC) alloy. Although crack initiation occurs in a given overlay or aluminide vane coating just as readily for CC, DS, or SC substrates, the crack growth rate is markedly reduced (6,7) for the DS and SC materials (Figure 4-2). The reason for this behavior is the low elastic modulus of the DS and SC alloys parallel to their growth direction. They grow in the (001) crystallographic direction, and the elastic modulus in this direction is 40 percent below that for CC alloys. As a consequence, a given thermal strain parallel to the growth direction results in stresses and stress intensity factors that are only 60 percent of those for CC alloys.

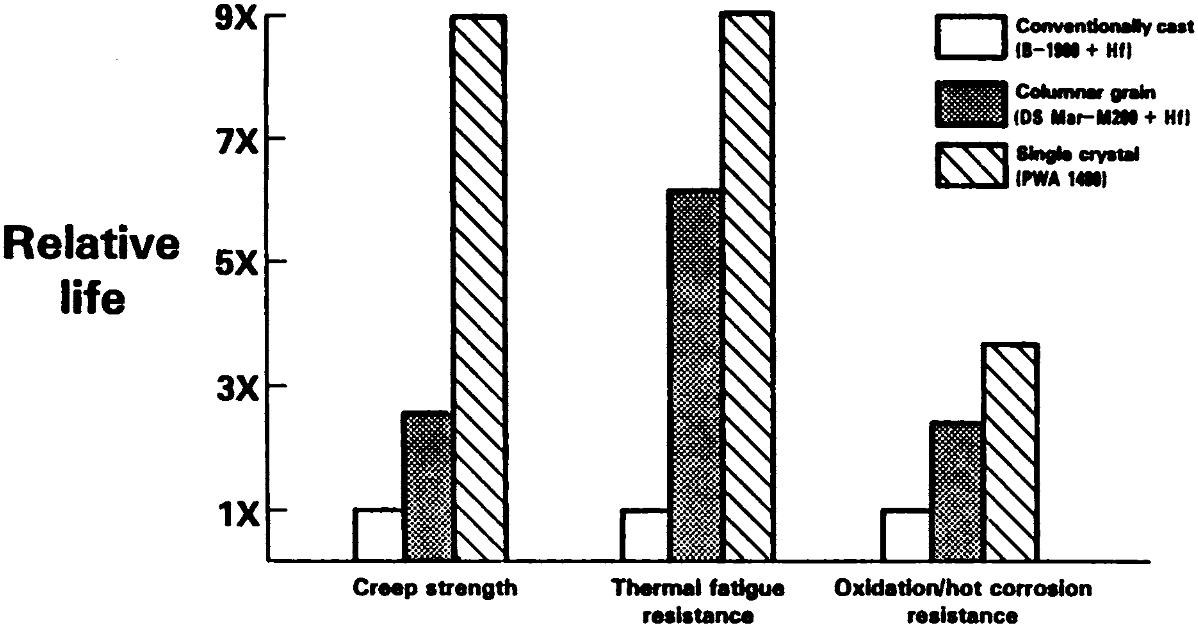


Figure 4-1. Relative comparison of several properties of state-of-the-art conventionally cast, columnar-grain, and single crystal superalloys.

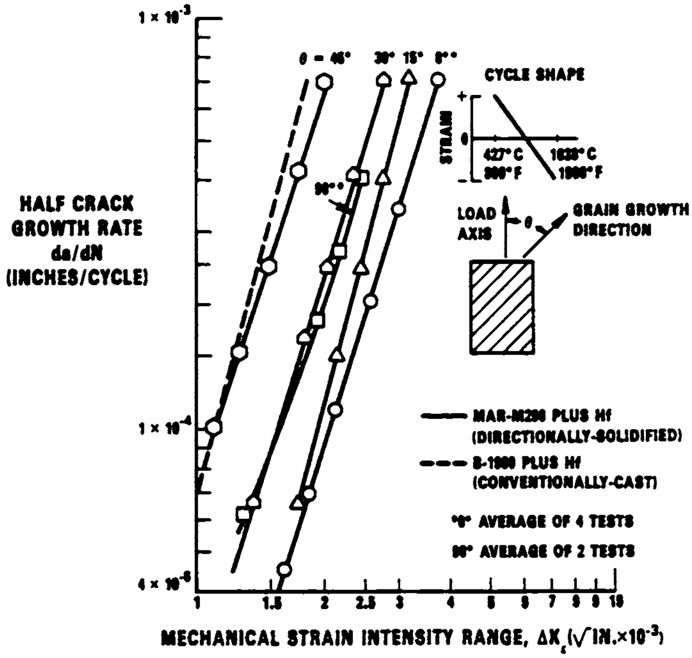


Figure 4-2. Thermal-mechanical fatigue crack growth rate for directionally solidified Mar-M200 + Hf at various orientations and for conventionally cast B-1900 + Hf.

In directions other than parallel to the growth direction, thermal stresses can be equivalent to or even somewhat higher than those experienced by CC alloys.

Under the sponsorship of the Naval Sea Systems Command, certain DS and SC alloys have been developed that have attractive combinations of creep-rupture strength and hot corrosion resistance for marine gas turbine applications. This combination of properties would also offer advantages for large land-based gas turbines. The materials include a DS alloy, N6202 (8), that has creep strength equivalent to and hot corrosion resistance 3 times that of IN 792, the most corrosion-resistant CC alloy. The single crystal alloys are Navaloy 300, 400, and 500 (9). Navaloy 300 has equivalent creep strength and 8 times the hot corrosion resistance of IN 792. Navaloy 400 shows a slight improvement in hot corrosion resistance and a 50°F increase in creep capability compared to IN 792. Finally, Navaloy 500 has equivalent hot corrosion resistance but a 100°F increase in creep-rupture strength compared to IN 792.

Certain conditions must be met to successfully cast columnar grain or single crystal components (10). A temperature gradient must be obtained such that supercooling of the liquid metal in advance of the solid-liquid interface is never encountered. This precludes nucleation of extraneous grains. In addition, heat flow must be unidirectional such that dendritic growth can only occur parallel to the temperature gradient. In practice these conditions are realized by setting the molds on a water-cooled chill plate, heating the mold and molten metal by induction methods, and withdrawing the mold from the hot zone in a controlled manner (Figure 4-3).

In the case of relatively small aircraft gas turbine hardware, multiple components (6 to 24) are grown simultaneously in the same mold. This can be achieved because of the relatively small total volume of metal involved and because of the use of baffling, which restricts lateral heat losses. Casting vendors have little experience, however, in fabricating columnar grain and single crystal components in the sizes required for large land-based gas turbines. In some instances, existing facilities may be incapable of handling large castings. Even if they can handle them, there are unanswered questions regarding their ability to grow large single crystal blades. For example, could single crystallinity be maintained? Could multiple parts be cast simultaneously, or would the parts have to be cast one at a time? Would the economics of having to grow one at a time be acceptable? Clearly, much effort is needed to assess the limits of existing casting technology and, possibly, improve on this technology for large land-based gas turbine applications.

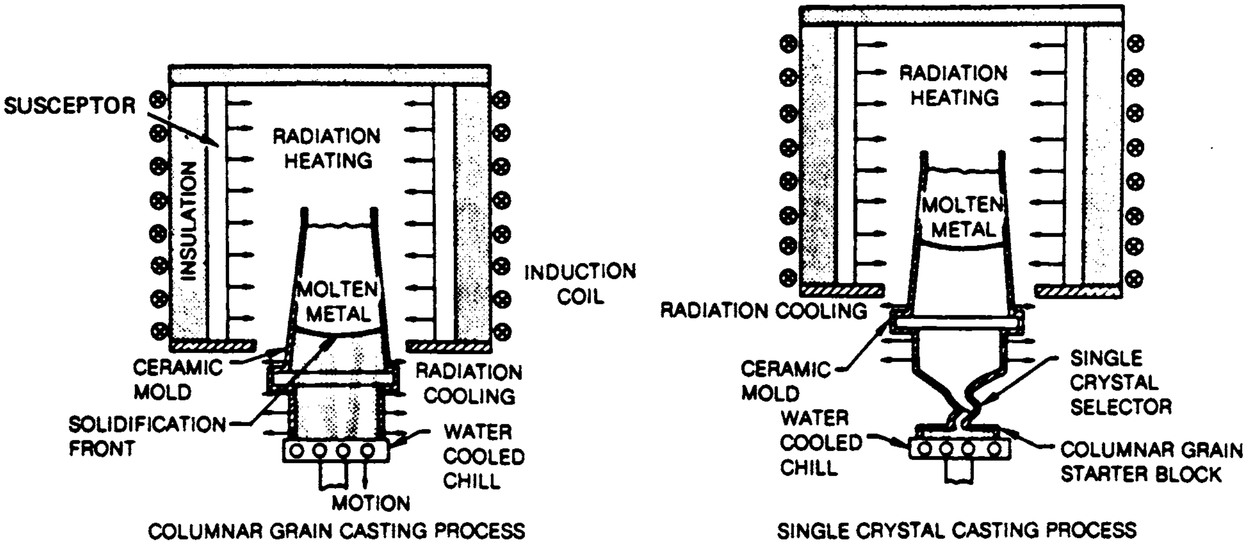


Figure 4-3. Directional solidification and single crystal casting.

The following conclusions can thus be drawn regarding directionally solidified and single crystal superalloys:

- Substantial improvements in creep-rupture resistance and thermal fatigue crack growth resistance can be achieved by replacing conventionally cast nickel-base superalloy turbine blades with either columnar-grained or single crystal blades. Single crystal alloys offer an enhanced combination of mechanical properties and hot corrosion and oxidation resistance compared to available columnar-grained alloys.

- Existing or recently developed alloy compositions are very likely to be suitable for large land-based gas turbine applications.

- Little casting development work has been conducted on large-size DS and SC components.

Accordingly, the committee makes the following recommendation:

- The feasibility of economically producing large-scale DS and SC components should be addressed through an experiment program combined with a comprehensive cost/benefit study.

RAPID SOLIDIFICATION TECHNOLOGY AND POWDER METALLURGY

Rapid solidification technology (RST) as used here refers to that practice in which alloys are produced by cooling from the liquid at rates much higher that those encountered in ingot metallurgy. In line with the objectives of this study, we consider here primarily those instances of RST that are intended to achieve improved mechanical, wear, or environmental resistance in a bulk product, as opposed to activities such as melt extraction for low-cost wire or flake production, production of fine aluminum powder for use as such, or direct rolling of aluminum into sheet or foil for cost reduction. Powder metallurgy (PM) undertaken to afford metallurgical advantage from the higher solidification rates involved is included, but that directed toward cost reduction alone is not. Thus qualified, RST and PM will be used interchangeably in the following discussion. A reason for interest in RST and PM for future land-based gas turbines is an anticipated need for high-performance turbine disks larger than those used in aircraft gas turbines.

This segment examines the advantages of using RST and PM to make parts of the equipment in large land-based gas turbines. Current status of the technology is reviewed briefly, and the conclusions reached indicate that, although technical advantages may be attainable, economic constraints are likely to inhibit many applications.

The potential advantages of RST (11) are

- Compositional uniformity, avoiding gross segregation;
- Uniformity and fine distribution of precipitates and second phases;
- An expanded range of possible alloy compositions;
- Possible achievement of a very fine grain size;
- Attractive mechanical properties resulting from the above;
- Attractive wear or environmental properties resulting from the above; and
- The advantages, where they obtain, of powder metallurgy practice.

Starting about 1970, high-speed tool steels produced by powder metallurgy have been in full-scale commercial production and are now standard products available from several sources. Gas atomization is used to produce the alloy powder, which is then consolidated to full density or consolidated and hot-worked to full density. Two superalloys have also reached full-scale production status via RST-PM, namely, low-carbon IN-100 and Rene 95. Both are used for aircraft gas turbine parts, including disks, the former in Pratt and Whitney and the latter in General Electric Company engines. Comparative ultimate tensile strengths of disk alloys are shown in Figure 4-4.

Two new high-strength aluminum alloys, 7090 and 7091, which use powder metallurgy to avoid segregation, have also reached at least the pilot-plant production stage.

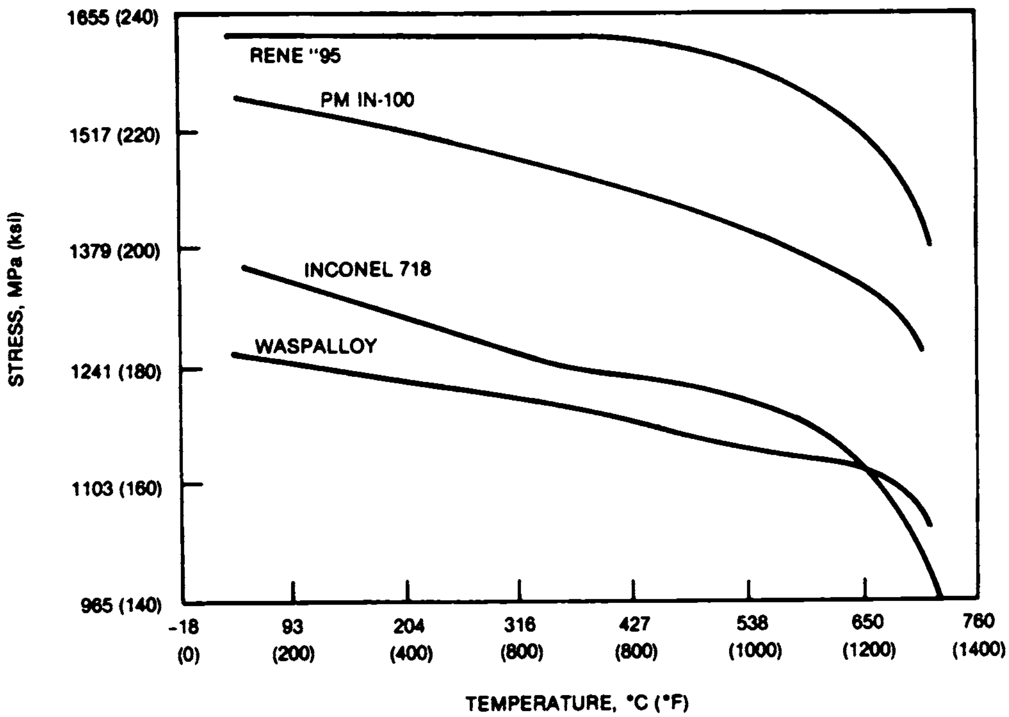


Figure 4-4. Comparison of ultimate strength of Rene 95 versus other disk alloys. The alloy compostions are as follows (12):

Alloy	C	Cr	Ni	Co	Mo	W	Cb	Al	Ti	Fe	Other
Rene 95	0.06	13	Bal	8.0	3.5	3.5	3.5	3.5	2.5	--	
IN-100	0.07	12	Bal	18.5	3.0	--	--	5.5	4.3	--	
IN-718	0.1 max	18	Bal	--	3.0	--	5.2	0.6	0.8	18	
Waspaloy	0.10	19.5	Bal	13.5	4.0	--	--	1.3	3.0	0.75	0.0045, 0.06 Zr

Brazing foil in the Metglas alloy series and a high-vanadium wear-resistant alloy are also in production.

A vigorous government program for development in the RST area is now being prosecuted at an expenditure level of about $25 million per year. Alloys for gas turbine hot parts, high-temperature and high-rigidity aluminum alloys, and alloys for high-performance, long-life gears and bearings are included in development efforts. As a result of these efforts, additional alloys that outperform currently available materials should be expected.

The probability that materials produced via RST and PM will find advantageous application in large land-based gas turbines will be influenced very strongly by economic factors. These will vary with the types of parts concerned. The RST-PM alloy will, except in unusual cases, be more expensive than alloys produced by ingot metallurgy because of the additional processing steps involved. The cost penalty will vary from quite modest, as in some tool steels where increased production yields largely offset increased PM cost, to very appreciable, as would be an attempt to make large single crystal turbine blades by starting with fine powder and recrystallizing directionally in the solid state to form a single crystal. Directionally solidified or single crystal blades produced by casting would be expected to win on economic grounds against such competition.

Although PM alloys are now being used for turbine disks in military aircraft gas turbines, the prospects for using PM alloys as disks in large land-based gas turbines have increased but are not yet very high. It is noted that in a recent new design for a large commercial aircraft gas turbine a manufacturer well experienced in PM disks elected to use an ingot metallurgy alloy. Optimistic forecasts of the advantages of parts hot-isostatically pressed to shape have been common, but there is at present a reluctance on the part of aircraft gas turbine manufacturers to use such parts in critical highly stressed applications such as turbine disks. For very large parts the size constraints of available extrusion or hot-isostatic pressing equipment may also impose processing constraints. The sizes of production-scale hot isostatic presses currently available in the United States are shown in Table 4-1. Notwithstanding such factors, the ability of RST-PM processing to avoid segregation, which increases with ingot size and alloy complexity, and to afford a fine grain size, which frequently persists in heat treatment, remains. It would thus not be too surprising to see a PM disk in a future large land-based gas turbine.

For applications in land-based gas turbines, the prospects for RST-PM products are more likely. Wear-resistant alloys or alloys with exceptional environmental resistance in compositions not amendable to ingot practice, gear or bearing alloys improved via RST microstructure control, or alloys with special characteristics (e.g., damping) producible only by RST may well be competitive in an environment demanding very long, trouble-free life.

Therefore, the following conclusions and recommendations are made regarding RST and PM:

- Conclusion--Rapid solidification technology and powder metallurgy should offer a series of attractive technical advantages in construction of advanced large land-based gas turbines, and a

Table 4-1

PRODUCTION-SCALE HOT ISOSTATIC PRESSING FACILITIES (13)

Organization	Working Hot-Zone Diam. x Length (in.)		Temperature Maximum (°C)	(°F)	Pressure Maximum (psi)
Battelle	19	52	1500	2730	15,000
Cameron Powder	27	65	1230	2250	15,000
Crucible	54	120	1260	2300	15,000
Howmet	20	69	1230	2250	25,000
	41	96	1260	2300	17,400
IMT	16	60	1450	2650	15,000
	16.5	63	1450	2650	15,000
	17	56	1230	2250	15,000
	38	96	1260	2300	15,000
	41	48	1260	2300	15,000
Kennametal	15	100	1400	2550	15,000
Pratt and Whitney	48	84	1260	2300	15,000
Udimet AE	30	65	1500	2730	15,000
Wyman-Gordon	47	60	1260	2300	15,000

vigorous government-supported program for development of these technologies is now in progress.

- Recommendation--The land-based gas turbine community should monitor progress in the development efforts to identify, as early as possible, potentially attractive new materials and fabrication techniques.

- Conclusion--Single crystal technology appears attractive in very-high-temperature, high-stress applications (e.g., turbine blading), but single crystals are difficult and expensive to produce from RST-PM starting materials.

- Recommendation--Although all development areas should be watched, it would probably be unwise for the large land-based turbine community to make significant investments in RST-PM alloys intended for single crystal applications in view of a probable loss in competition with a casting approach.

- Conclusion--A series of alloys with attractive properties such as wear resistance, environmental resistance, and high damping have been produced by RST-PM, and development efforts are continuing in these areas.

- Recommendation--The large land-based gas turbine community should be particularly alert in seeking RST-PM alloy developments that may offer cost advantages via longer life and trouble-free operation in component parts. Attention would be focused primarily on auxiliary equipment rather than on the main components of the gas turbine itself.

OXIDE-DISPERSION-STRENGTHENED ALLOYS

The design and development of advanced gas turbine engines is governed by the ability to increase the turbine inlet temperature. Increasing this temperature permits gains in the thrust-to-weight ratio and reduction in specific fuel consumption. These gains are frequently achieved by design improvements in the turbine section, which includes sophisticated cooling schemes for blades and vanes. The use of cooling air results in a performance penalty caused by aerodynamic losses as well as increased costs for fabrication of components containing intricate cooling passages. Oxide- dispersion-strengthened (ODS) alloys may provide a means to use uncooled blades and vanes at higher turbine inlet temperatures. Furthermore, when used with cooling, ODS alloys may permit operation at higher metal surface temperatures.

International Nickel Company (14) developed the mechanical alloying process shown in Figure 4-5 to produce ODS alloys. Typical of this process, mixtures of elemental metal powders, master alloy powders, and very fine refractory oxide particles are prepared. These mixtures are charged into a high-energy ball mill or attritor, where mechanical alloying is obtained. The individual particles are cold-welded and fractured by repeated high-energy mechanical impact until the alloy consists of a uniform dispersion of a highly refractory oxide (e.g., Y_2O_3) in a metallic superalloy matrix.

The mechanically alloyed powders are then consolidated by canning, vacuum degassing, and hot extrusion, followed by hot-working operations and subsequent annealing to a highly textured microstructure. A final solution-annealing followed by an appropriate aging treatment is then used to optimize the strength properties.

Tien (15) has summarized some of the effects of particles on a number of materials properties relevant to alloy design for structural applications. He concluded that oxide dispersions do have utility to enhance flow strength and creep resistance, especially at high temperatures. However, adverse effects are usually observed in regard to uniaxial ductility and toughness.

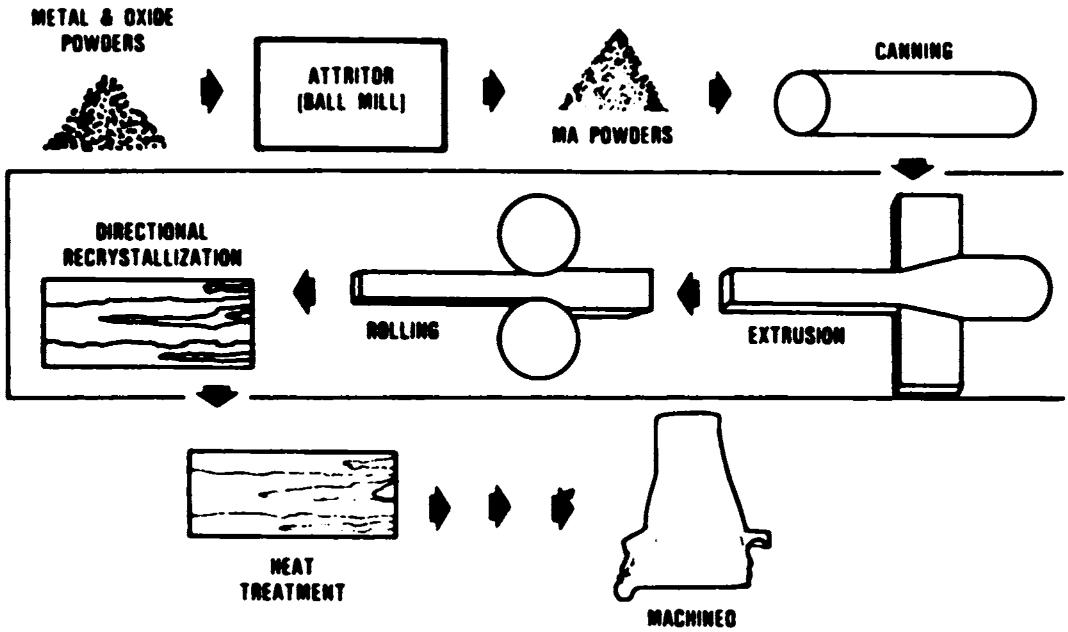


Figure 4-5. Schematic of the mechanical alloying process for the production of high-pressure turbine blades.

Beatty and Millan (16) have compared a number of the properties of ODS superalloy MA6000 and DS MAR-M 247. Table 4-2 compares the compositions and physical properties of MA6000 and DS MAR-M 247 in terms of density, thermal conductivity, mean coefficient of thermal expansion, and dynamic modulus of elasticity. The lower density of MA6000 compared to DS MAR-M 247 is advantageous for blade application, since a lighter blade results in lower disk rim stresses. The thermal expansion and thermal conductivity of the two alloys were similar, but the high modulus of MA6000 was considered to be unfavorable since peak load stresses can be generated by smaller transient thermal strains.

The tensile properties of the two alloys as a function of temperature are compared in Figure 4-6. The ultimate tensile strength of MA6000 is higher with lower tensile elongation (Figure 4-7) compared to DS MAR-M 247 up to 1300°F. From 1300°F to 1800°F, the cast alloy surpasses MA6000 in tensile strength because of its higher volume fraction of gamma-prime. Above 1800°F, however, the tensile strength of MA6000 did not decrease as rapidly as DS MAR-M 247 because the oxide-dispersion-strengthening mechanism was still active.

Average creep and stress-rupture strengths for both alloys were also compared (Figures 4-8 and 4-9) in the longitudinal orientation using conventional Larson-Miller parameter presentation with an arbitrary constant of 20. These results

Table 4-2

COMPOSITION AND PHYSICAL PROPERTIES OF MA6000 AND MAR-M 247

Characteristic	MA6000	MAR-M 247
Composition (percent)		
Ni	Bal	Bal
Cr	15	8.25
Co	--	10
W	4	10
Al	4.5	5.5
Ti	2.5	1
Fe	2.5	--
Ta	2	3
Mo	2	0.7
Hf	--	1.5
C	0.05	0.15
B	0.01	0.015
Zr	0.15	0.09
Y_2O_3	2.5 v/o	--
Density (lb/in.^{3})	0.292	0.312
Thermal conductivity (Btu/hr-°F)		
1000°F	9.50	8.75
1400°F	11.25	10.75
1800°F	13.25	13.00
Coefficient of thermal expansion (X 10^{-6} in./in./°F)		
1000°F	6.6	7.2
1400°F	7.4	7.6
1800°F	8.5	8.5
Dynamic modulus of elasticity (X 10^6 psi)		
1000°F	27.2	16.0
1400°F	24.4	15.0
1800°F	18.0	12.4

showed that DS MAR-M 247 had higher creep and stress-rupture strengths than MA6000 at lower temperature and higher stresses, but the ODS alloy became superior at higher temperatures and lower stresses because of the oxide-dispersion-strengthening mechanism.

Oxidation and hot corrosion tests were also performed on these alloys as well as on specimens of IN 792 and IN 738. MA6000 was comparable in oxidation resistance to the cast alloys after 400 hours at 2000°F, whereas its hot corrosion resistance was superior to these same alloys after 250 hours at 1700°F in a burner rig using 5 ppm sea salt. These oxidation and hot corrosion tests cannot be used to examine the

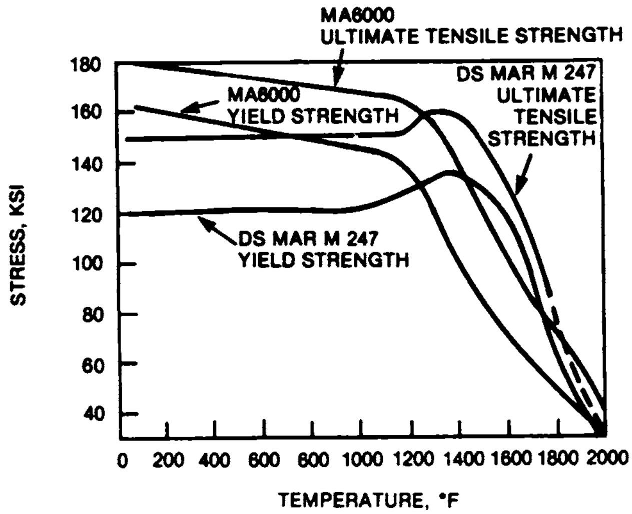


Figure 4-6. Tensile properties versus temperature of MA6000 and MAR-M 247 (longitudinal orientation).

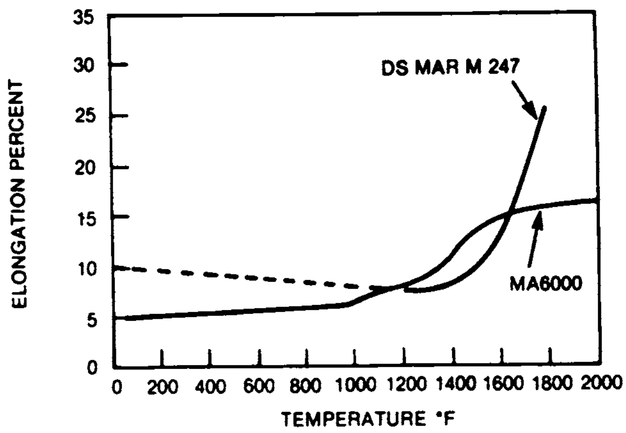


Figure 4-7. Elongation versus temperature of MA6000 and MAR-M 247 (longitudinal orientation).

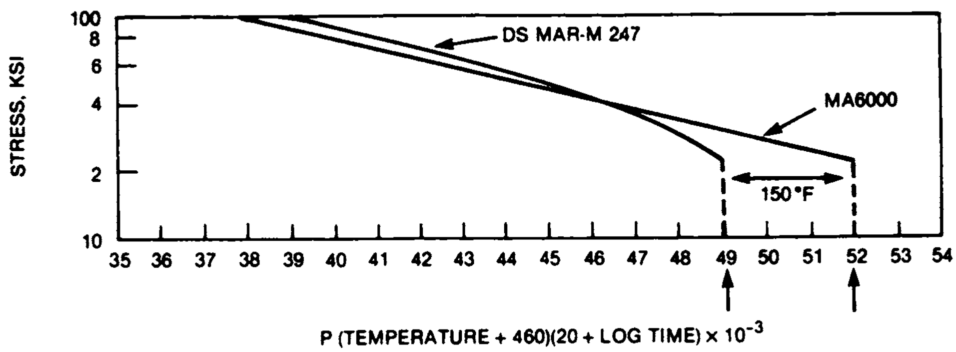


Figure 4-8. One percent creep data for MA6000 and MAR-M 247 (longitudinal direction).

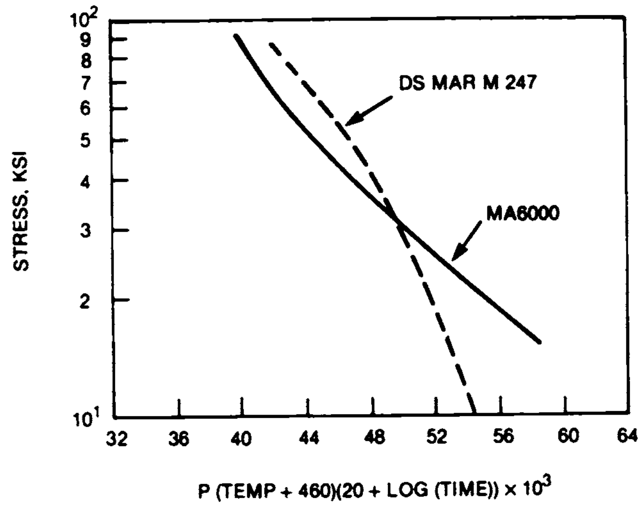


Figure 4-9. Stress-rupture data for MA6000 and MAR-M 247 (longitudinal orientation).

effects of the oxide dispersion since significant compositional differences were present in the alloys that were compared. Moreover, these alloys probably would not be used in the uncoated condition.

The comparison of MA6000 and DS MAR-M 247 has given encouraging results with regard to using an ODS alloy for uncooled blade applications. More testing is required. ODS alloys are not now widely used in gas turbines, but their use is becoming more widespread. Initial applications have been as combustion burner liners.

The following conclusion and recommendation are made regarding ODS alloys:

- Conclusion--ODS alloys have the potential for use in the combustion section as well as in uncooled turbine hardware in applications that would require air cooling of conventionally cast superalloys. It may also be possible to use ODS alloys with cooled hardware to operate at higher temperatures. At present, however, ODS alloys are not standard bill of materials for aircraft gas turbines.

- Recommendation--It is recommended that ODS alloys should not be considered as a prime material for the design of advanced land-based gas turbines. However, the use of this type of alloy in aircraft gas turbines should be closely monitored, and ODS alloys may have utility in the design of advanced land-based gas turbines when additional experience with these alloys becomes available from aircraft gas turbines.

ORDERED ALLOYS

This segment examines the potential of ordered alloys in the construction of large land-based gas turbines. Current status of the technology is reviewed briefly, and the conclusions reached indicate that, although it is too early to make reliable predictions regarding eventual success, there is adequate reason for the land-based gas turbine community to follow, and perhaps to participate in, the development work.

Ordered alloys are alloys in which there are two or more atomic species and these two species occupy specific sites in the crystal lattice. Such alloys tend to occur at well-defined atomic ratios (i.e., AB, A_3B, AB_3, etc.) A large number of ordered alloys have been known for many years. They have characteristically exhibited low ductilities and, as a result, are encountered predominantly in specialty applications such as superconductors (e.g., Nb_3Sn), magnetic materials (e.g., FeSe), thin coatings (e.g., NiAl), or constituent phases of structural alloys (e.g., $CuAl_2$, Ni_3Al) rather than as structural alloys per se. However, this view of the appropriate limits of applicability of ordered alloys has changed in recent years as the result of work that has shown that ductilities can in some cases be dramatically improved by proper application of metallurgical principles. The National Materials Advisory Board has reported recently (17) on the current status of ductile ordered alloys and their application potential, and the views expressed here reflect, in large measure, the information accumulated in that effort.

Recent development efforts in the United States (Table 4-3) have concentrated attention on ordered alloys in the aluminides of iron, titanium, and nickel. These alloys characteristically exhibit low densities, high moduli, high work-hardening rates, and low steady-state creep rates compared with conventional alloys.

Strengths frequently increase with temperature until a critical temperature is reached and, even at this stage in development, specimens may be competitive on a density-corrected basis with conventional high-temperature alloys. Figure 4-10 compares the yield strengths of advanced long-range ordered alloys and nickel aluminides with several commercial structural alloys.

Thus far, although laboratory test results are very promising and development work in titanium aluminides is well advanced, no ordered alloy has reached the state of development at which sufficient manufacturing and design data have been generated to permit engineering utilization of such alloys in other than experimental hardware. However, the mechanical properties being observed, particularly at elevated temperature, combined with a high aluminum content, which promises good oxidation resistance at elevated temperatures, direct attention to the aluminides as potentially very useful alloys for future gas turbine design. It should be possible to achieve a competitive balance of mechanical and environmental properties in such alloys at a

Table 4-3

RESEARCH EFFORTS ON ORDERED ALLOYS IN THE UNITED STATES (17)

Laboratory	Source of Funding	Systems Studied	Principal Investigators
NASA Lewis	NASA	FeAl	D. Whittenberger
AFWAL	AF	Fe_3Al, TiAl, Ti_3Al	H. Lipsitt
Systems Res. Lab.	AF	Fe_3Al, FeAl, Fe_3Si	M. G. Mendiratta
Gen. Elec. R&D Center	AF	Ni_3Al+B	S. C. Huang
Pratt and Whitney			
E. Hartford	--	Ni_3Al	D. Duhl
W. Palm Beach	AF	Fe_3Al+B, TiAl	E. Slaughter
E. Hartford	AF	TiAl	M. Blackburn
Oak Ridge National Lab.	DOE	$(FeNi)_3V$, Ni_3Al+B	C. T. Liu
	ONR	Fe_3Al	C. T. Liu
Olin Corp.	Internal	Fe_3Al	--
Marko Materials	DARPA/AMMRC	FeAl+B	R. V. Ray
Univ. of Pennsylvania	NSF	Ni_3Al	D. P. Pope
	DOE-ORNL	$(FeNi)_3V$	D. P. Pope
Dartmouth Univ.	DOE-ORNL	Ni_3Al, Fe_3Al	E. Schulson
	NASA	NiAl	E. Schulson
Vanderbilt Univ.	DOE-ORNL	$(FeNi)_3V$, Ni_3Al	J. J. Wert
Case Western Reserve Univ.	AF	Fe_3Al+B	K. Vedula
	NASA	FeAl, NiAl	K. Vedula
	NSF	NiAl	R. Gibala
Stanford Univ.	NASA	CoAl, NiAl	W. Nix
Texas A&M Univ.	NASA	NiAl, CoAl, FeAl	A. Wolfenden
Rensselaer Polytech. Inst.	DOE-ORNL	$(FeNi)_3V$, Ni_3Al+B	N. S. Stoloff
	ONR	Fe_3Al, Ni_3Al	N. S. Stoloff
Northwestern Univ.	NSF	Au-Ni	J. B. Cohen

significant weight saving over conventional alloys. This would make them attractive in aeronautical construction and provides the basis for the interest in the aeronautical community in their development.

In large land-based gas turbines, sizes will be predominantly larger and the advantage of weight saving should be less than in aeronautical equipment. The factor of cost, although important in both cases, should be more critical in the commercial equipment. The ability to make large, high-quality parts at low cost using ordered alloys will be critical to their successful application in future large land-based gas turbines. The prospects for success in doing so are open to question at present because of the current lack of information and experience in this area. Although, with minor exceptions (e.g., Hf in Ni_3Al), the constituent elements of the ordered alloys are not more expensive than those of the superalloys

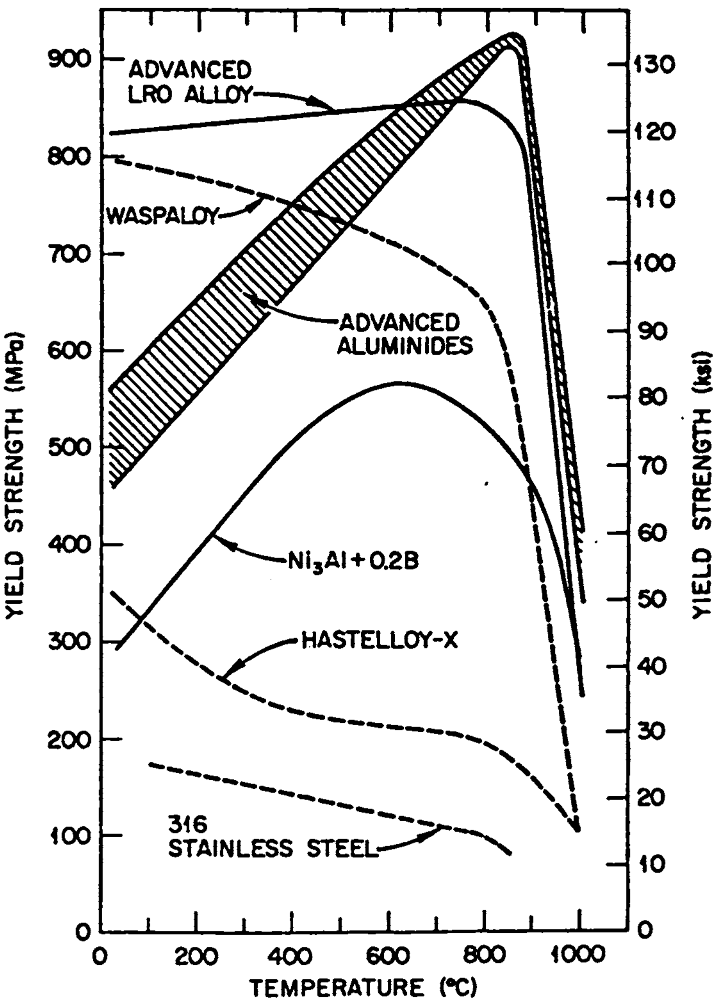


Figure 4-10. Comparison of the yield strengths of advanced long-range ordered alloy and nickel aluminides with commercial structural alloys. The yield strength of ordered intermetallic alloys increases with temperature. (Source: Unpublished material by C. T. Liu, Oak Ridge National Laboratory, 1983.)

now in use in large land-based gas turbines and it appears thus far that available melting and processing equipment will be adequate, there are factors that may lead inexorably to higher cost for ordered alloys. It is known, for example, that small changes in stoichiometry can cause major property changes in some ordered alloys, implying a need for more stringent composition control than is normal for superalloys. Impurity content and grain size control may also be more critical.

Although the final outcome of the competition between ordered alloys and the more conventional alloys for use in large land-based gas turbines cannot yet be predicted, it would seem advisable for the large land-based gas turbine community to maintain contact and familiarity with development efforts to facilitate transition of successful developments. Accordingly, the following conclusion and recommendation are reached with regard to ordered alloys:

- Conclusion--Ductile ordered alloys in the iron, nickel, titanium, and cobalt aluminide series constitute a class of materials that, based on developmental results now being obtained, appear attractive as potential materials for gas turbine engine construction.
- Recommendation--The large land-based gas turbine community should monitor progress in development efforts to identify, as early as possible, potentially attractive new materials.

DUAL ALLOY ROTOR TECHNOLOGY

The acronym DART as originally used in a DARPA-sponsored effort stood for dual alloy radial turbine, but it is here broadened to signify dual alloy rotor technology. DART processing refers to a deliberate tailoring of the composition of a part to achieve optimum properties in the various sections. Thus, in a gas turbine rotor the sections near the hub might be made in a composition and grain size yielding highest burst strength at modest temperatures, whereas the higher temperature areas near the rim could have a different composition and grain size optimized for creep or stress rupture strength. This would be achieved by powder metallurgy practice and thermal or thermomechanical processing.

With respect to large land-based gas turbines, DART processing may be considered an extension of rapid solidification and powder metallurgy technology, and the observations made on that topic should apply here also. DART practice is not now in production. However, a small-diameter radial turbine rotor has been made successfully. In view of the added complications that would be encountered in much larger parts, it would appear advisable to let DART technology mature, as it may, on smaller equipment before according it significant attention for possible application in large land-based gas turbines.

COMPOSITE MATERIALS

There are a great variety of filament-reinforced composite materials available or under intensive development that may be of value for advanced large land-based gas turbines. Although the history of composites can be traced for centuries, it is primarily in the past 25 years that reinforcing filaments of sufficient strength and stiffness have become available to produce composites that exceed the properties of metallic alloys. The filaments of interest include compositions from all three major classes of materials and in turn are utilized to reinforce members of all three classes--namely, ceramics, metals, and resins.

Various alternative choices of desirable properties can be made with filament-reinforced composites. For example, the specific strength and stiffness (i.e., tensile strength or modulus of elasticity divided by density) in the uniaxial case can be up to an order of magnitude greater than the values for the matrix material by itself.

The benefits of using composites are very significant in the appropriate applications. For example, the use of graphite epoxy composite skin and structure in modern military aircraft to save 25 to 30 percent of the weight of the metal replaced is already at about 25 percent, and it is projected that up to 50 to 60 percent of future aircraft structures will be based on filament-reinforced composites. Much of these benefits are due to the weight-saving attributes of composites. These are of less interest or value to land-based gas turbines, so any use of composites in such machines will necessarily be based on other attributes. Among the other potentially useful properties available in some composites are ease of fabrication, especially of large pieces, lower cost, damping of both acoustical and vibration energy, and ease of maintenance, largely due to their corrosion resistance. Some of these benefits will be mentioned as appropriate in the following discussion of each type of composite.

In this segment, all three types of composites as defined by the matrix materials (resins, ceramics, and metals) are discussed.

Resin Matrix Composites

Organic resins, generally reinforced with glass, graphite, organic, and occasionally other kinds of filaments, are by far the best developed and most widely used advanced composites. Both military and commercial aircraft airframes and engines as well as spacecraft, missiles, and land-based vehicles are utilizing ever-increasing amounts of advanced filament-reinforced resins.

In the case of gas turbine aircraft engines, some 5 or 6 percent of their current weight is organic composites. This is expected to grow to 15 percent over the next few years. Weight savings of 20 to 50 percent (compared to the metal parts being replaced) are being obtained, along with both direct and indirect cost savings. In the case of the C5A aircraft engines, for example, 390 metallic components per engine were replaced with 200 lb of plastic composites in each of the four engines, which at 50 percent weight savings saved a total of 800 lb with a direct cost saving due to increased payload and/or flight range worth several million dollars over the 30,000-hour expected life of the engines (18). These components are primarily in cool portions of the engine, although organic resins with higher and higher use temperatures are being developed. For example, a graphite fiber-reinforced polymer outer compressor duct for the F404 engine to operate at -65°F to 485°F in steady-state environments and up to 550°F transients has recently been announced. It may replace a chemically milled titanium duct that had already been weight- and cost-optimized. The composite is expected to save another 15 percent in weight and cost $9,000 less than the titanium component (19).

Land-based gas turbines, however, do not have the same impetus to save weight as aircraft engines or the high-value materials against which to compare costs. Nevertheless, it appears fruitful to continue to consider at least the lower cost fiberglass-reinforced resins for application to land-based gas turbines. Among the potential benefits would be corrosion resistance, high damping capability, and lower life-cycle costs in such areas as the inlet and filter ducts and auxiliary housings. These materials are already utilized to some degree, but increased applications are apparently warranted at this time.

Future applications of advanced composites may be justified as costs decline and the experience with them in aircraft engines warrants their adaptation to land-based gas turbines. Thus it can be concluded that

- The organic resins appear to offer lower life-cycle costs along with other benefits in many applications. They should be carefully considered for applications in and around land-based gas turbines where corrosion, maintenance, noise, and appearance are important considerations.
- High-performance advanced resin composites are being widely used in aircraft gas turbines and should be considered as costs and experience indicate potential usefulness on land-based gas turbines.

Ceramic Matrix Composites

Ceramic matrix composites, consisting of a ceramic matrix with dispersed ceramic fibers or particles, or possibly both, have attracted increasing attention because of the favorable changes in mechanical behavior they commonly produce. These changes are of particular interest because they reflect direct or implied improvements in mechanical reliability. While there is, indeed, significant promise and some demonstration of improvements in the mechanical behavior of such composites, there are important issues that suggest caution in considering them for application in large land-based turbines. Since both the promise and the concerns of ceramic matrix composites have recently been reviewed in detail (20), this report will only briefly outline the advantages and concerns.

A variety of mechanisms have been identified or postulated as the sources of change in mechanical behavior of the ceramic composites (21) from that of normal monolithic ceramics. One reason that none of the mechanisms is understood in all details and some are only partially understood is that most, if not all, ceramic composites have at least two, if not more, mechanisms operating simultaneously. While this adds to the complexity and uncertainty of these materials, it also provides substantial opportunity for improving or tailoring resultant composites.

Overall, there are two basic types of composites: those in which the dispersed phase is in the form of particulates and those in which the dispersed phase is in the form of fibers, usually continuous fibers. This distinction in type of composite is important, not only from the standpoint of constituents and processing that can be used, but also from the standpoint of behavior. Specifically, the particulate composites will exhibit a truly, or very nearly truly, linear load deflection curve to the ultimate fracture stress, at which point they will typically fail catastrophically as a "normal monolithic ceramic" will; in contrast, a continuous fiber composite will typically show substantial load-carrying capability at strains beyond those at which such composites exhibit their maximum load-carrying capability (Figure 4-11). While one can clearly conceive of the possibility of

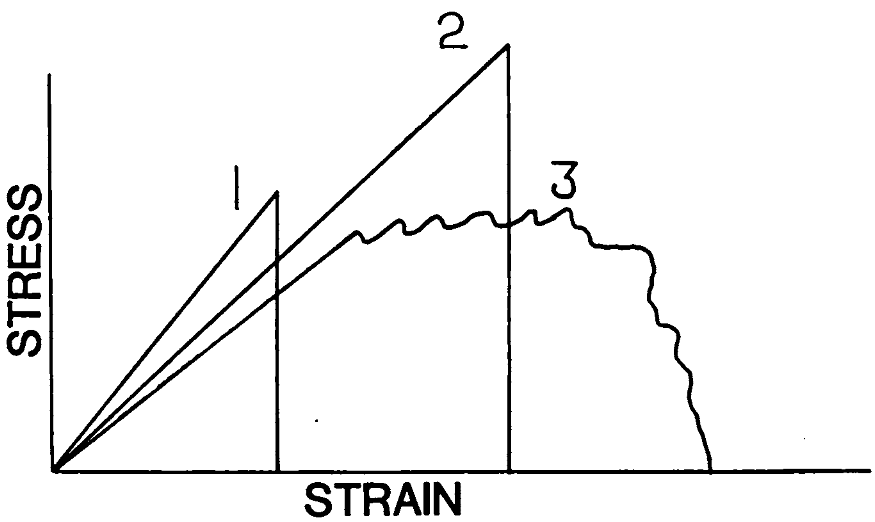


Figure 4-11. Approximate stress-strain behavior of (1) monolithic ceramics, (2) a particulate-reinforced ceramic, and (3) a fiber-reinforced ceramic.

combining fiber and particulate composites, the presence of continuous fibers should dominate the behavior so that the combination should behave, primarily, as a fiber composite. There may be possibilities of using short fibers, or whiskers, to achieve intermediate types of behavior between those described. However, results to date have shown that composites containing whiskers exhibit catastrophic failure like particulate composites and traditional ceramics.

While particulate (and whisker) composites thus characteristically show catastrophic failure like a "normal monolithic ceramic," there are a variety of mechanical differences between them and normal ceramics. These differences, in turn, depend on the one or more mechanisms that are important in their behavior. Thus, where we have transformation toughening from the use of zirconia particles in a metastable tetragonal state, substantially higher strengths than achieved in typical monolithic ceramics can be obtained. In contrast, most, if not all, other mechanisms involved in particulate composites, including such concepts as microcracking, crack deflection, and crack branching, may give strengths comparable to those achieved in the matrix by itself as a pure monolithic material, but they also commonly result in some reductions in strength. On the other hand, continuous fiber composites can result in strengths similar to, and in some cases substantially above, those that would be achieved in the matrix by itself as a monolithic material.

The key differences that exist between these composites in their room-temperature mechanical behavior and normal monolithic ceramics is in their resistance to fracture. Particulate composites will commonly exhibit improvements in the strain to failure (Figure 4-11) under normal loading and/or show improvements in other modes of failure, such as under thermal or mechanical shock (e.g., due to particle impact). On the other hand, the substantial load-carrying capability beyond the maximum load-carrying capability of typical continuous fiber composites clearly shows a very large change in resistance to failure. It is, in fact, these changes in resistance to failure, especially in the ceramic fiber composites, that have attracted so much attention. One particular type of continuous fiber-reinforced ceramic composite under development and not yet thoroughly studied or characterized in detail has exhibited outstanding performance in a year-long furnace test. It is discussed later in this section.

There are, however, important uncertainties of the extent and scope of these benefits. The first is their performance at elevated temperatures, where there may be either benefits or deterrents, depending on the particular composite system as well as its use. The benefits of dispersed particles in a matrix typically will decrease with increasing temperature, in most cases disappearing at temperatures where sufficient diffusion or deformation mechanisms are available for stress relief. Transformation toughening disappears typically even before these temperatures, i.e., when the transformation temperature has been reached. However, since transformation-toughened materials also typically have at least one other mechanism operative, it has been demonstrated in single crystals that transformation toughening can show remarkable strengths--e.g., 100,000 psi to quite high temperatures (at least 1500°C) (22). Nevertheless, it appears likely that grain boundary sliding will occur, and may often be enhanced, in transformation-toughened materials as well as in composites utilizing dispersed particles to provide crack deflection, crack branching, or microcracking. This raises important questions about the extent of their application at high temperatures, especially for sustained periods.

Ceramic fiber composites, at least those utilizing glass matrixes, have also been shown to provide benefits in terms of mechanical performance at elevated temperatures relative to that of the matrix alone; that is, the fiber "reinforcement" of glass matrixes has shown significant improvement in their high-temperature strength and creep resistance. On the other hand, ceramic fiber composites, at least those showing good strength characteristics, have generally shown a disturbing propensity for embrittlement caused by high-temperature exposure. This appears to stem from the fact that toughness in these composites is associated with poor bonding between the fiber and the matrix. The fibers of primary use, because of their ability to withstand the processing additions, are the SiC-based Nicalon fibers, which undergo oxidation at high temperatures. It is thus believed that high-temperature oxidation of the surface of such fibers provides a mechanism for forming strong bonding between the fiber and the matrix, which thus embrittles the system upon cooling back to room temperature. What makes this of even greater concern is the fact that in ceramic matrix composites the matrix, as opposed to the fibers, is the low-strain-to-failure element. Thus the matrix begins microcracking before the composite fails. Therefore, even in a fully dense ceramic fiber composite, microcracking will occur well below the failure stress, and this can expose the material to initiation of the environmental embrittlement phenomena. Similar embrittlement phenomena may occur in oxide matrixes containing oxide fibers caused by sintering phenomena or possibly caused by combustion impurities infiltrating into the composite and providing fiber matrix bonding.

Another major concern for composites, primarily if not exclusively those involving some degree of microcracking (which thus includes most particulate and all fiber composites), is that they show mechanical fatigue. This is clearly exhibited both as thermal shock fatigue (i.e., degradation of the thermal shock resistance of the material with repeated exposures to substantial changes in thermal stress) and as true mechanical fatigue (i.e., progressive reduction in the ultimate load-carrying capability due to repeated mechanical loading), even in the absence of environmental effects, which can apparently further exacerbate these mechanical fatigue effects.

These serious issues of environmental embrittlement and mechanical fatigue appear to be at least partially, and possible even totally, controllable. However, much work remains to determine to what extent this can be accomplished, in what systems, and under what operational conditions. In view of the long life and high levels of reliability demanded in large land-based turbines, these issues must be taken as a very serious caution in considering either type of ceramic composite in such systems at this time.

There is also another related set of basic issues--fabrication and cost. Most, and quite probably all, ceramic particulate composites, at present and for the foreseeable if not indefinite future, face the same basic size and shape strength issues associated with conventional monolithic ceramics. Thus, in view of the substantial size and frequent complexity of shapes of hot-stage components in large land-based turbines, there is similar uncertainty for the use of particulate composites in these applications as for the use of normal monolithic ceramics. In principal, some fiber composites (e.g., continuous fiber composites) could be fabricated by combinations of such techniques as filament winding and hot isostatic processing that could potentially provide the size, shape, and quality of component needed for turbine applications. However, there is substantial uncertainty whether this can truly be accomplished and, if so, over what time frame and at what cost.

A possible exception to at least some of the concerns and problems discussed here may be achievable through the use of continuous glassy or glass-ceramic refractory fibers and a chemically vapor-deposited or infiltrated matrix. Numerous samples of these constructions, ranging from simple tubular shapes in sizes up to 8 in. diameter and 8 ft long to coils and flanged and closed-end tubes, have been fabricated and are being evaluated for various applications. Test results for applications that include static gas turbine components and furnace tubes suggest that continuing consideration be given to potential applications in large land-based gas turbines as cross-flame or cross-fire tubes, combustors, and transition pieces.

Among the factors that may have permitted the achievement of some encouraging test results with these particular materials are the low processing temperatures, the low modulus of the continuous fiber phase, and the high-density thin wall of the composite, which permits high heat-transfer rates with small thermal gradients. There are undoubtedly other factors that may either help or hinder the performance of these composites, but preliminary results of tests are sufficiently encouraging to warrant continued effort.

- Conclusion--Ceramic composites offer important changes in mechanical behavior that could make them more desirable materials for use in large land-based gas turbines than present materials.
- Conclusion--These composites also have critical issues associated with them--in particular, mechanical fatigue for almost all such composites and oxidation or environmental embrittlement in ceramic fiber com- posites--that must be much more clearly understood and effectively addressed before such composites could be incorporated into turbines. Substantial process- ing advances would have to be achieved to make this practical.
- Recommendation--Work to address these mechanical and processing issues should be undertaken to assess whether the intriguing changes in mechanical behavior of ceramic composites over normal mono- lithic ceramics can be effectively applied.
- Recommendation--Some of the mechanisms, especially those in particulate composites, should be studied for utilization as guidelines in terms of further improving ceramic coatings, which are also an important type of candidate material for potentially nearer term use in large land-based turbines.

Metal Matrix Composites

Refractory filament-superalloy composites are potentially useful in the most demanding turbine applications. Successful development and application of such composites to turbine blades could permit blade use temperatures as high as 2000 to 2100°F without diffusion-barrier-coated wire and as high as 2300°F with diffusion barriers (23-26).

Composites offer a promising approach for achieving structural materials that combine the high-temperature strength of refractory materials while preserving the oxidation resistance, toughness, and flaw tolerance of metals. These composites include ceramic-ceramic, ceramic-metal, and metal-metal combinations. The ceramic-ceramic composites achieve improvements in toughness and reliability by interrupting cracks at the weakly bonded fiber-matrix interface. However, weak bonds limit the strength contribution of the higher modulus, higher strength fiber phase. Ceramic-metal and metal-metal composites use a ductile matrix to bond together strong filaments into useful structural forms. Matrix deformation augmented by filament disbonding provides fracture toughness for this type of composite. Refractory metal

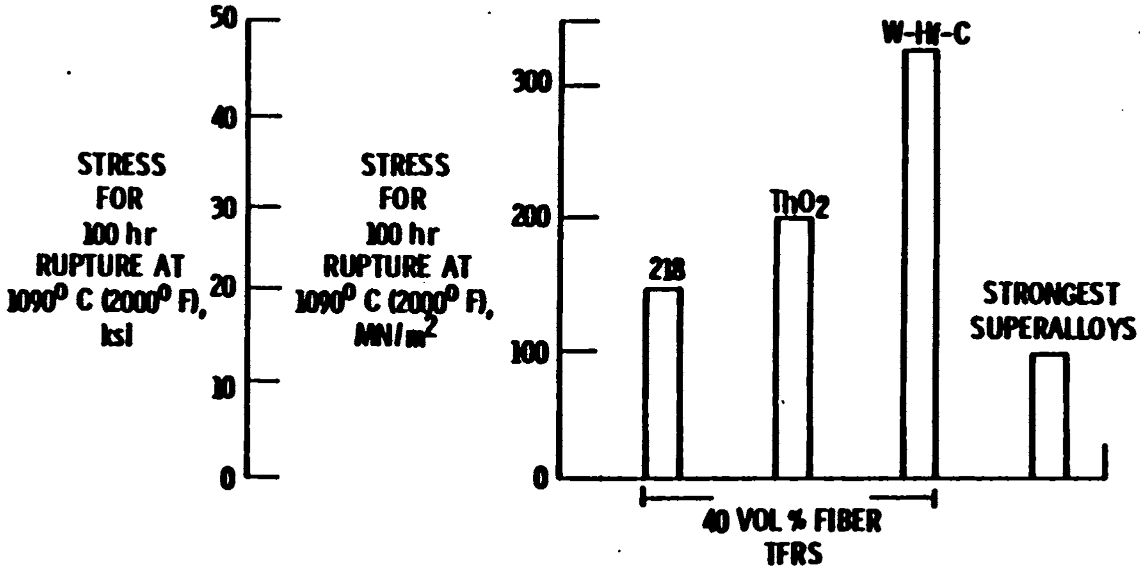


Figure 4-12. Tungsten filament reinforced superalloy composite rupture strength at 1090°C (2000°F).

wire offers less potential gain in high-temperature strength but more plastic deformation and toughness compared with ceramic filaments. Further, the property degradation for refractory metal wire is a time- and temperature-dependent gradual loss rather than the relatively sudden loss associated with ceramic filament-reinforced metals. Although several combinations of filaments and metal matrixes are under development, most are intended for low- to moderate-temperature applications. The greatest effort and highest temperature potential are embodied in the tungsten wire-superalloy composites.

Refractory wire-superalloy composites have been studied for aircraft turbines. The rupture strength shown in Figure 4-12 compares composites with superalloys for that application. The low filament content and weak matrix are not well suited for land-based turbine requirements. Weak matrix alloys were chosen to resist thermal-cycle-induced stress while providing the creep strength for aircraft life cycles with more stringent weight limits. Another advantage that tungsten wire composites have over conventional superalloys and ceramic materials is a higher thermal conductivity. Tungsten has 3 times the thermal conductivity of nickel-, iron-, and cobalt-base superalloys. Thus, a tungsten wire composite has 1-1/2 to 2 times the thermal conductivity, which is effective in dissipating thermal transient temperature gradients and improving cooling. This is helpful in designing vanes to reduce thermal-gradient-induced stresses (27). The advantages of a first-generation refractory filament-reinforced superalloy are as follows:

- Coatings unnecessary up to 1100°C
- Fiber-matrix reaction negligible to 1100°C
- Low creep rates at 1100°C
- Thermal conductivity over 1.5 times that of superalloys
- Good impact strength
- High and low cycle fatigue resistance
- Thermal fatigue resistance

The fabrication of a complex, hollow, air-cooled airfoil of tungsten filament-reinforced superalloy (TFRS) was undertaken using simple modifications of the process developed previously for aluminum composite fan and compressor blades. This process is shown schematically in Figure 4-13. A first-stage JT9D turbine blade was fabricated from TFRS using diffusion bonding of monolayer TFRS composite plies along with steel core plies and unreinforced cover skin plies at the inner and outer surfaces of the blade. After diffusion bonding, the steel was leached from the airfoil, leaving a hollow configuration. An impingement insert could be inserted to improve the interior cooling airflow path. The effort to fabricate the blade was successful, demonstrating the feasibility of fabricating TFRS components with

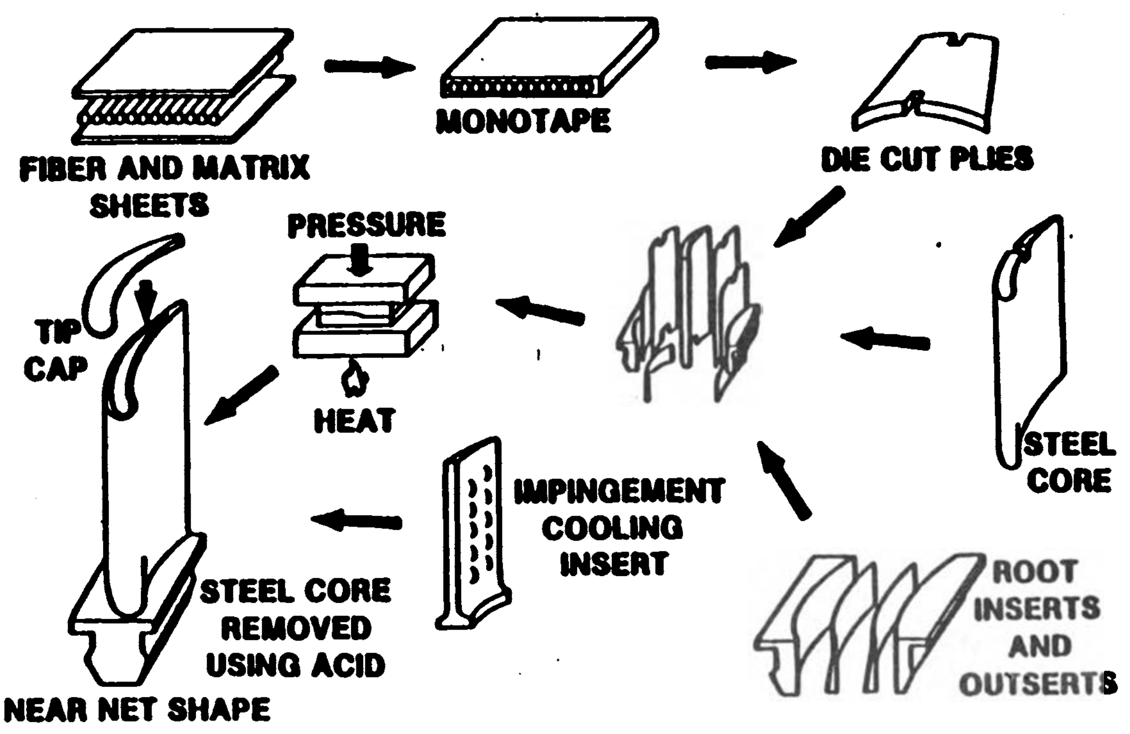


Figure 4-13. Schematic of TFRS fabrication process.

complex geometries using the methods proposed (28). Although considerable effort to evolve the production processing technology for TFRS remains, fabrication feasibility for producing complex components has been demonstrated, and the results achieved can serve as the basis for that development.

Several operational factors facilitate application of refractory wire-superalloy composites to a power-generating turbine when compared with an aircraft turbine. Thermal cycling associated with shutdown and startup is infrequent for land-based turbines in contrast to aircraft operation, and thus thermal cycle fatigue failure is much less important as a failure mode. Further, thermal barrier coatings and oxidation and corrosion protective coatings are less likely to fail from thermal-cycle-induced spalling. Also, there is much less restriction on blade weight, since increased turbine disk size and weight incur much less of a penalty for land-based turbines. The high weight limit means that higher filament content can be used to increase strength, since filament content for aircraft turbine blades is kept low to limit blade and disk weight. Tungsten wire has demonstrated almost an order of magnitude specific-strength advantage over conventional superalloys for 1000 hours rupture at 2000°F. Higher filament contents can increase strength/density values to better provide the long service life needs for land-based power turbines. Disk and blade attachment design trade-offs to accommodate a higher blade weight need to be addressed.

Tailoring of properties to match the application requirements is one of the advantages cited for composites. As an example, for aircraft turbine use, the 3000 start-stop cycles per year dictate that the matrix for a refractory wire composite be ductile even at the sacrifice of strength in order to accommodate the plastic strain resulting from the thermal cycles (29). For base power gas turbines, the thermal cycle constraints on matrix alloy selection are of much less concern. However, the very long operating service life of up to 100,000 or more hours requires a much higher creep strength and resistance to reaction at the matrix-fiber interface. Filament strength loss from interdiffusion and reactions at the interface are strongly dependent on matrix chemistry (30).

Fiber-matrix reaction and degradation have been studied for times over 1000 hours but not for the very long times required for land-based turbines. Also, diffusion barriers for wires have been explored as a means to increase use temperature without loss from interdiffusion and reaction but again not for the long times necessary for land-based turbines. Barrier coatings of carbides and oxides were effective but difficult to produce without isolated defects. Barrier coatings and stronger matrix alloys are a necessary area of research for refractory wire-superalloy composites for land-based gas turbines to promote the long-time creep strength. Matrix plasticity can be reduced somewhat, but chemical compatibility of the matrix with a barrier-coated filament needs research effort.

While fabrication of airfoils has been demonstrated for aircraft turbines, fabrication of the large blades for land-based gas turbines has not been explored. The fabrication process developed for aircraft turbines has been modified and improved recently (31) whereby an arc-spray technique was used to produce a single-layer fiber-matrix sheet or monolayer. Sheets 38 by 122 cm have been made, and larger sizes can be readily produced, which should satisfy the size requirements for land-based gas turbines. Hot isostatic pressing of the arc spray monotape has been used to consolidate stacked monolayers into the appropriate composite geometry. Use of HIP diffusion-bonding rather than die hot-pressing increases the ease and reduces the cost for larger components, since the die and press capacity limitations are circumvented. Large HIP facilities are available that can process multiple-blade batches. The fabrication process is applicable, but the specific parameters must be developed and demonstrated.

The following conclusions and recommendation regarding tungsten wire-reinforced superalloys are thus warranted:

- Conclusion--Refractory wire-superalloy composites offer the potential for improved creep rupture strength, high cycle fatigue resistance, and more effective heat transfer that could permit increased operating temperature compared with the best superalloys.

- Conclusion--Matrix alloy chemistry and wire diffusion barrier coatings are promising areas for study to further increase properties for long-time service.

- Conclusion--Increased fiber content composites also can increase properties for a longer service life, but design trade-off studies for disk and blade attachment geometries are required.

- Conclusion--Fabrication processing for the larger blade sizes appears straightforward and economically feasible but must be demonstrated.

- Conclusions--Significant development effort is needed before tungsten wire-superalloy composites can be applied to large land-based gas turbines.

- Recommendation--The large land-based gas turbine community should monitor progress in the development of tungsten wire-superalloys to permit their expeditious utilization in appropriate applications should the opportunity arise.

COATINGS

This segment discusses the coatings technology that is available for use in an advanced industrial gas turbine where filtration of the intake air and clean fuels will be used. Turbine inlet temperatures up to 2600°F (1427°C) are anticipated.

As described in Section 3, "State of the Art of Materials for Large Land-Based Gas Turbines," the principal needs for coatings in current land-based gas turbines are these:

- For compressor airfoils, coatings may be used to develop aqueous corrosion resistance and to control surface roughness to maintain performance.
- For the combustion system, combustion baskets and transition ducts require coatings to develop resistance against high-temperature oxidation.
- For stationary turbine vanes and rotating turbine blades, coatings are required to develop oxidation resistance but especially to obtain resistance to Type II hot corrosion attack.

An advanced gas turbine using filters and clean fuel and having a turbine inlet temperature of 2600°F may not have hot corrosion degradation via Type II attack. If clean air and fuel are used, hot corrosion resistance should not be an important design requirement. On the other hand, the alloys used must have some resistance to the various forms of hot corrosion attack, otherwise intermittent use of impure fuel (e.g., fuel with substantial amounts of alkali metals, vanadium, and sulfur) or lack of adequate intake air filtration could result in severe attack of turbine blades and vanes.

Although some resistance to hot corrosion conditions is necessary, the high turbine inlet temperature and clean air and fuel indicate that the principal cause of environmentally induced alloy degradation should be high-temperature oxidation. Experience with aircraft gas turbines operating under conditions proposed for the advanced industrial gas turbine can be used to determine the type of coatings that should be used in order to obtain resistance against high-temperature oxidation.

In the compressor section the coatings currently available for land-based gas turbines should be adequate. For example, nickel-cadmium or aluminum-rich sacrificial coatings may be used. If nickel alloys or titanium alloys are used, as may be the case where compressed-air temperature exceeds 800°F (427°C), diffusion aluminide coatings should be considered in order to obtain adequate oxidation resistance.

In the combustion system, higher temperatures will be encountered in the advanced gas turbines than in current land-based gas turbines. Aircraft gas turbine experience shows that the comparatively high temperatures in the combustion system can be compensated for by using thermal barrier coatings such as yttria-stabilized zirconia. The technology for such coatings is now rapidly evolving and should be adequate for the advanced gas turbine.

The principal need for coatings in the advanced gas turbine will be in the turbine section to obtain resistance against oxidation. The operating conditions can be expected to be significantly different from current state-of-the-art land-based gas turbines. In particular, the operating temperatures will be higher, and hot corrosion should not be a significant problem. Aircraft experience can be used to select coatings, but there may be significant differences between aircraft gas turbine operating conditions and those in advanced land-based gas turbines. For example, in aircraft gas turbines, coatings have been developed not only to obtain oxidation resistance but also to obtain resistance against thermal fatigue. The thermal fatigue problem arises from the gas turbine engine cycle, and it may not be as much of a problem in the advanced land-based gas turbine as it is in current aircraft gas turbines. This would be the case especially if the advanced land-based gas turbine is used for base-load as opposed to peak-load operations.

Based on aircraft gas turbine experience, a variety of coatings are available to obtain resistance against oxidation and hot corrosion. All of these coatings have been developed by attempting to have a reaction product develop on the surface of the coating that inhibits all subsequent reaction between the coating and the environment. The most effective reaction-product barriers against oxidation and hot corrosion are Al_2O_3, Cr_2O_3 or SiO_2 (32). The fabrication and properties of coatings available for high-temperature applications have been discussed in a number of papers and reports (32-39). A convenient means to discuss these high-temperature protective coatings is by considering the fabrication process, which can include diffusion, physical vapor deposition, and plasma spray.

A great variety of diffusion coating processes are available to coat alloys. They may involve aluminizing, chromizing, or siliconizing or adding some other metal to the surfaces of alloys. All of these processes must include a means to bring the metal to the surface of the alloy to be coated and conditions for this element to diffuse into the alloy. The metal for coating can be provided by plating, dipping, painting, or applying it in vapor form. Diffusion into the alloy is achieved by use of a sufficiently high temperature. Although a great number of techniques are available to fabricate diffusion coatings, the variety becomes even greater because within a given technique more subtle variations are possible. For example, in the case of aluminide coatings fabricated by using the pack cementation process, the source of aluminum (e.g., pure aluminum or alloyed aluminum powder), the type of activator (e.g., NH_4Cl, NH_4Br, or NH_4F), and the size of the alumina filler material all have profound effects on thickness and composition of the coating. Furthermore, in some processes, additional steps may be taken to add additional elements to the coatings (e.g., platinum, yttrium, or hafnium).

It is sufficient here to propose that, since diffusion coatings are relatively cheap in comparison to other types of coatings, such coatings should be considered for use in advanced gas turbines. Numerous aluminide or duplex aluminide modifications are available. The final selections will depend on the proposed turbine metal temperatures.

Overlay coatings are not as dependent on substrate compositions as diffusion coatings. Overlay coatings can be fabricated by a number of different techniques, with the two principal methods being physical vapor deposition and plasma spray. Both of these procedures are line-of-sight processes and are therefore not practical for internal passages. The physical vapor deposition process frequently is accomplished by electron beam melting to form a vapor cloud or by sputtering. The plasma spray process must involve gas shielding or low pressures, since otherwise oxidation of elements in the coatings such as yttrium and aluminum becomes a problem.

The microstructures of coatings are different when fabricated by physical vapor deposition or by plasma spray. In the case of electron beam physical vapor deposition, the coatings contain leaders that must be sealed by peening and a suitable heat treatment. Plasma spray coatings are more equiaxed than PVD coatings, and the oxide content of plasma sprayed coatings is greater. In fact, preference is usually given to vacuum plasma spray (VPS) over other types of plasma spray coatings because of the lower oxygen content. Nevertheless, it appears that, even in coatings fabricated by VPS, elements with high affinities for oxygen and low concentrations in coatings are virtually completely converted to oxide.

A great number of overlay coating compositions have been developed for high-temperature use (Table 4-4). These coatings are frequently called the MCrAlY coatings since chromium, aluminum, and yttrium are usually always present and M may be Fe, Ni, or Co or a mixture of these elements. The concentration of the elements depends on the intended use of the coatings. For the case of hot corrosion,

Table 4-4

OVERLAY COATING COMPOSITIONS
("CoCrAlY")

Co	Cr	Al	Y	Comments
Bal	17	8	0.3	ATD-41
Bal	18	8	0.5	LCO-29
Bal	18	9	0.3	ATD-6
Bal	18	11	0.3	PWA 268C, ATC-30
Bal	18.5	7	0.3	ATD-6B
Bal	18.5	7.5	0.3	ATD-6C
Bal	18.5	8	0.3	ATD-6D
Bal	19	12	0.3	PWA 68J, ATD-4J
Bal	21	10	0.3	BC-21, ATD-2B
Bal	21	10	0.3	ATD-42
Bal	23	12	0.3	ATD-2C
Bal	25	11	--	1-4 Hf/Zr, ATD-18
Bal	26	10	--	ATD-30
Bal	28	12	0.3	ATD-12
Bal	31	5	0.3	BC-29, ATD-14
Bal	30	10	0.3	ATD-70
Bal	35	10	0.3	ATD-71

chromium is high (35 weight percent), whereas for oxidation resistance the chromium must be lower (20 percent) and aluminum should be kept high (11 percent). Other elements are now being added to MCrAlY coatings to attempt to obtain improved coating performance, including silicon, hafnium, tantalum (38) and platinum (39). The composition of the MCrAlY coatings to be used can only be selected after the conditions of use are specified, which includes alloy substrate composition. The useful lives of overlay coatings, and especially diffusion coatings, are markedly dependent on substrate composition, and it is usual practice to speak in terms of a coating system, which includes coating-substrate combinations, rather than individual coatings.

When gas turbines are operated at high metal temperatures, interdiffusion of the coating with the substrate usually plays a significant role in degradation of the coating. Coatings development for aircraft gas turbines currently is placing much emphasis on techniques to inhibit such interdiffusion. The lives of coatings to be used in the advanced gas turbine can be expected to be affected by interdiffusion with the substrates. Existing aircraft technology should be used, but research and development on procedures to inhibit intradiffusion between coatings and substrates may be required.

It is important to emphasize that a great number of coating systems are available for use in an advanced gas turbine. Many of the most recently developed compositions are proprietary, but numerous compositions are also available in the literature. As mentioned previously for diffusion coatings, some developmental studies on coatings behavior should accompany their use on a new turbine engine, since all new turbines exhibit certain unique operational characteristics. Otherwise, the coatings technology now available from aircraft and marine turbines appears adequate.

Ceramic coatings are used as thermal barriers to decrease metal part surface temperatures about 180°F. Such coatings are currently being studied with different cooling schemes to obtain better thermal barriers. The current technology consists of porous, stabilized zirconia coatings formed by using the plasma spray process. Stabilization is achieved by using Y_2O_3, CaO, or MgO, but other stabilizers are being examined. Thermal barrier coatings are currently used in the combustion section of some aircraft gas turbines. Work is in progress to attempt to use this type of coating on turbine vanes. To date, problems arising from oxidation of bond coats have not permitted the use of thermal barrier coatings on vanes in the turbine section.

Thermal barrier coatings are believed to be extremely important to the advanced land-based gas turbine. Such coatings will certainly be used in the combustion section, and very significant payoffs can be obtained if these coatings can be used for turbine blades and vanes. Research is therefore required on thermal barrier coatings in those areas not being adequately covered by the aircraft gas turbine industry. Since the heat rate of the advanced land-based gas turbine is different from the aircraft gas turbine, it is necessary to study the thickness of coatings appropriate for the land-based gas turbine. Other problems such as oxidation of the bond coat and the poor hot corrosion resistance of thermal barrier ceramic coatings must also be investigated.

Coatings are used to protect alloys from environmental degradation, but it is also necessary that the coatings used do not adversely affect the mechanical properties of the superalloy substrate. The effects of coatings on the mechanical properties of the substrates are not fully understood (40). There are examples where coatings produce beneficial as well as deleterious effects on the mechanical properties of the substrate. In the case of thermal fatigue, the coefficient of expansion of the coating is an important parameter and must be considered when the operating conditions are conducive to thermal fatigue failures. It appears that diffusion aluminide coatings are better than overlay coatings for applications where thermal fatigue is a problem, since cracks do not propagate as readily from a diffusion aluminide coating into the substrate as in the case of an overlay coating.

To satisfy environmental requirements, water probably will be injected into the combustion of the land-based gas turbine. The water vapor will therefore be significantly higher than in aircraft gas turbines. The presence of water vapor could affect the oxidation-induced degradation of coatings, but it is believed that this effect will not result in a significant difference in coating lives.

The following conclusions and recommendations regarding coatings are based on the foregoing discussion:

- Conclusion--The coatings required in the compressor sections of advanced land-based gas turbines should be essentially the same as those currently used in state-of-the-art industrial and aircraft turbines.

- Recommendation--The design of the advanced land-based gas turbine should rely on the existing technology for coatings used in the compressors of current state-of-the-art land-based and aircraft gas turbines.

- Conclusion--The combustion system in the advanced land-based gas turbine should not be subject to conditions any more severe than those encountered in state-of-the-art aircraft gas turbines. However, a direct scale-up from aircraft gas turbines may not be feasible.

- Recommendation--The design of the advanced land-based gas turbine should utilize the technology currently available for combustion system materials. In particular, use of thermal barrier coatings is recommended. Early in the design phase it will be necessary to determine if materials other than those used in aircraft gas turbines are required.

- Conclusion--The principal need for development of new coatings for the advanced land-based gas turbine is believed to be in the use of coatings for turbine blades and vanes. The coatings currently available for aircraft gas turbines may be adequate, but it is necessary to assess the need for resistance to thermal fatigue and hot corrosion in the coatings to be used. Moreover, the use of thermal barrier coatings may be especially appropriate for the advanced land-based gas turbine. Current aircraft gas turbine experience with such coatings is for much shorter times than the 100,000 to 200,000 hours of operation planned for the advanced land-based gas turbine, and therefore work is required not only to attempt to overcome the currently observed deficiencies in these coatings but also to determine their behavior during very long exposure times.

- Recommendation--The design of the advanced land-based gas turbine should use currently available coatings with high-temperature oxidation resistance and good diffusional stability. The exact compositions will depend on substrate compositions but probably will consist of a modified MCrAlY composition fabricated by low-pressure plasma spray or electron beam vapor deposition. Early in the design stage it will be necessary to assess the need for thermal fatigue resistance and hot corrosion resistance in the coating compositions. Thermal barrier coatings should be considered for use on turbine blades and vanes, but research and development efforts are required if the application of such coatings is to be successful.

GAS PATH SEALS

The most significant contribution to direct operating cost of commercial aviation gas turbine engines is fuel consumption, presently comprising more than 50 percent of the direct operating cost. Fuel consumption can be significantly reduced by improving component efficiencies throughout the engine. The efficiency of all of the major engine rotating components (compressors, turbines, fans) is very sensitive to clearances over the blade tips, at interstage seals, and flow through critical secondary airflow seal locations. It is estimated that for each 0.25 mm (0.01 in.) reduction in clearance realized throughout a typical commercial engine, a 1 percent standard fuel consumption savings would result, the equivalent of saving 50×10^6 gallons of fuel per year in the U.S. commercial aviation fleet.

In a modern commercial aircraft gas turbine engine there is a multitude of gas path seal locations, as shown in Figure 4-14, all of which are significant with respect

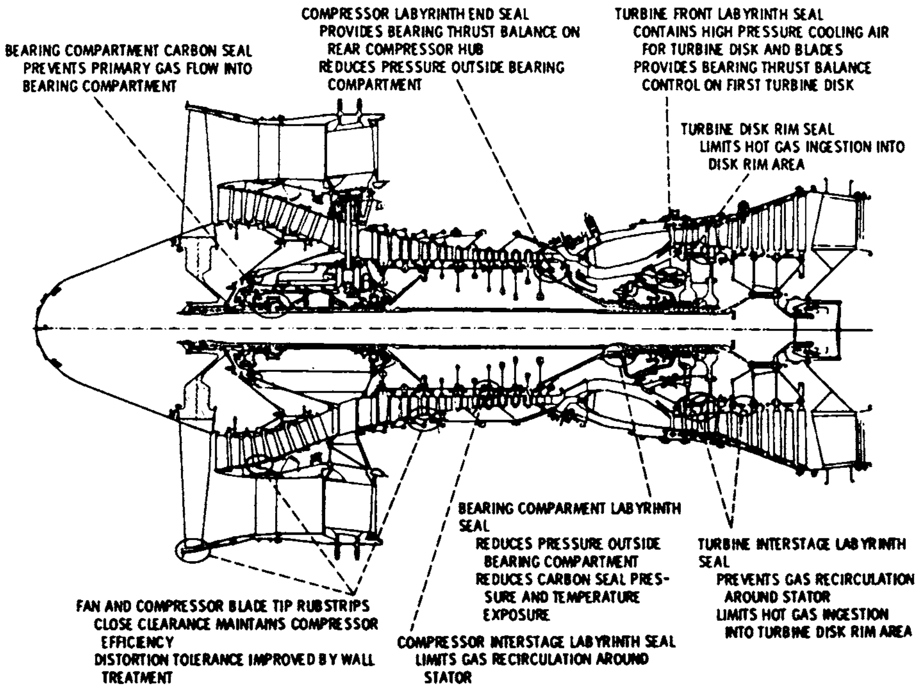


Figure 4-14. Modern transport engine.

to the performance and reliability of the engine. Among the gas path seals are the numerous labyrinth seals designed to reduce loss of high-pressure gas from the engine cycle, to control cooling air flow through the hot section of the engine, and to maintain pressure balance on the rotor shaft system. Also included in the gas path seal positions are the important outer gas path seal locations over compressor and turbine blade tips. The outer gas path seals are intended to maintain close operating clearance between the rotating blade tips and the stationary seal components.

A key to effecting reduced clearances is the successful application of abradable materials throughout the engine. The biggest payoff is in the high-pressure turbine, where significant cooling air reduction and clearance reductions have a very strong impact on improving engine performance. In this brief survey, the general state of the art of abradable seal technology is discussed, with a special focus on advanced turbine seal development.

Examples of materials used in both secondary gas path seal locations and in outer gas path seal locations are summarized in Figure 4-15. The primary consideration in selection of the seal materials for various locations is the local operating temperature. The desirable wear characteristic of each of the materials indicated in

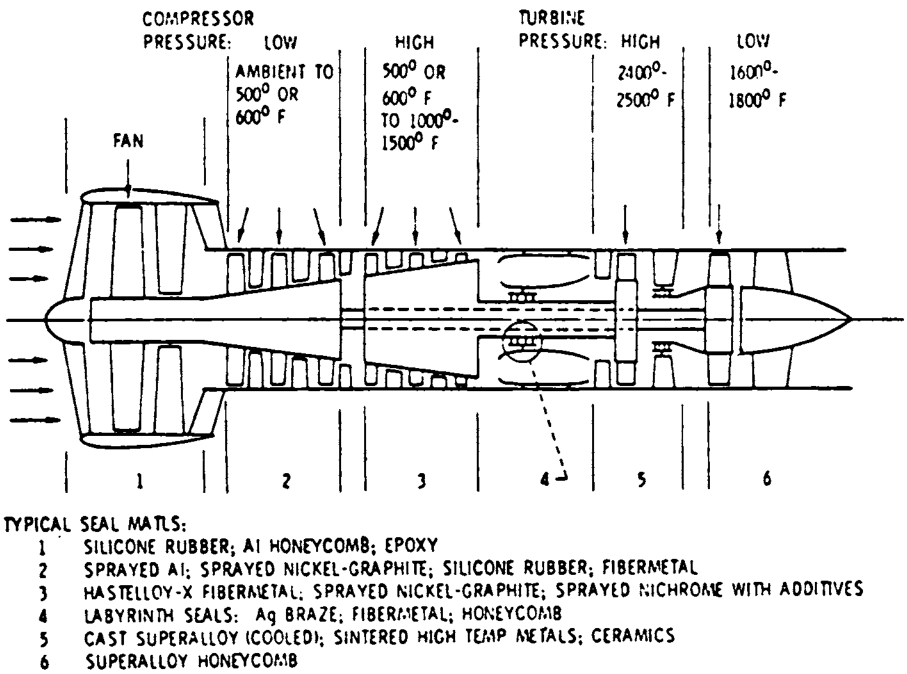


Figure 4-15. Summary of engine operating conditions and typical abradable seal materials.

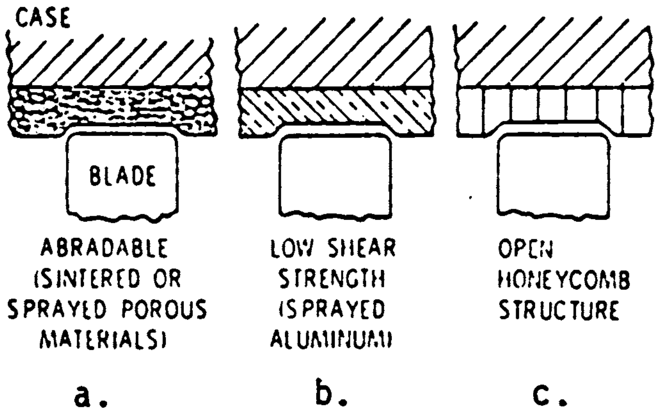


Figure 4-16. Illustration of types of abradable seal materials.

Figure 4-15 in their respective locations is that the seal material should wear rather than the rotating labyrinth seal knife edge or blade tip. In general there are three approaches (Figure 4-16) that may be followed to provide a gas path seal material with this desirable wear characteristic, often (somewhat misleadingly) called abradability.

The first approach, shown in Figure 4-16 (a), is to employ a highly porous low-density material. Such materials are usually prepared by sintering metal powders or fibers, often mixed with transient filler materials. It is also possible to plasma-spray such a structure by spraying metal particles mixed with easily volatilizable polymeric particles or with graphite. The desired wear behavior is afforded by the easy removal of discrete particles from the bulk seal material; fracture can readily occur across the small interparticle bond area when a rub occurs. Limitations inherent in this type of material include susceptibility to erosion damage and inefficient sealing due to leakage through open porosity.

The second class of gas path seal materials, shown in Figure 4-16 (b), includes denser structures (less than 30 percent porosity) that are often plasma-sprayed and in some cases sintered, hot-pressed, or even cast. Rub interactions for this class of seal materials are accommodated in a more complex manner than were those for the first type of seal material. Usually a combination of plastic deformation, material compliance (densification), and machining mechanisms is involved. Again, variations of this type of material are used in gas path seal locations throughout the engine: elastomeric materials are employed in low-pressure compressor positions, low-temperature metals in higher temperature compressor stages, and high-temperature materials in the turbine.

The third class of gas path seal materials, shown in Figure 4-16 (c), derives its rub tolerance from the geometric arrangement of the thin metal sheets from which the seal is fabricated. Probably the most widely used example of this type is the metallic honeycomb. The honeycomb cell walls are oriented in the radial direction. Consequently very little metal-to-metal contact surface area is involved when a rub interaction occurs. Honeycomb structures are generally applied in low-pressure turbine seal positions.

The single most significant gas path seal location is the high-pressure turbine blade outer air seal. Here, typically, 0.5 percent standard fuel consumption reduction can be realized for every 0.25 mm (0.01 in.) reduction in operating tip clearance. Here also the material challenges are the greatest, with gas temperatures of about 1500°C (2730°F). Seal materials must exhibit oxidation, erosion, and thermal shock resistance as well as provide a desirable rub surface. Ceramic material systems have recently been developed to meet these challenges and have been introduced in a commercial aircraft engine, with a significant benefit in terms of reduced cooling air requirements. Cooling air reduction benefits are summarized in Figure 4-17. In comparison to previous state-of-the-art MCrAlY materials (1100°C (2010°F) maximum surface temperature), a 1 to 2 percent cooling air savings can be realized if a ceramic material capable of 1300 to 1400°C (2370 to 2550°F) is employed.

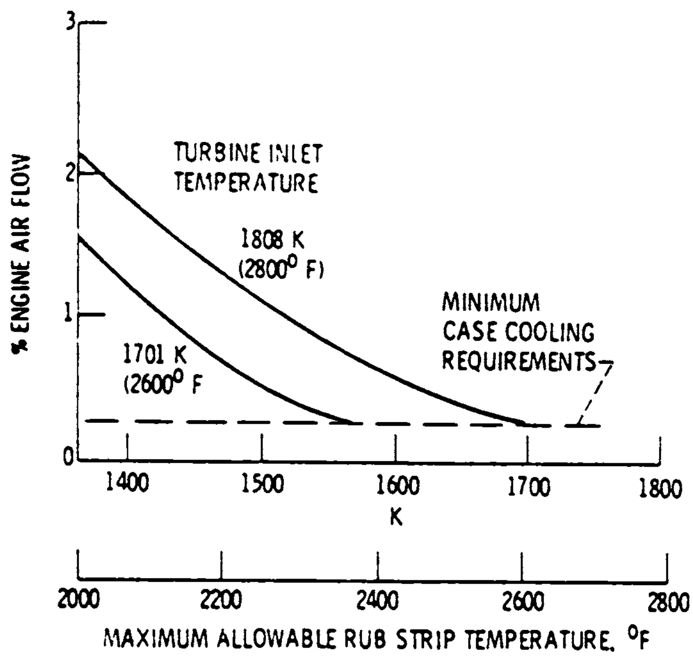


Figure 4-17. Cooling air flow requirement for high-pressure turbine outer air seal.

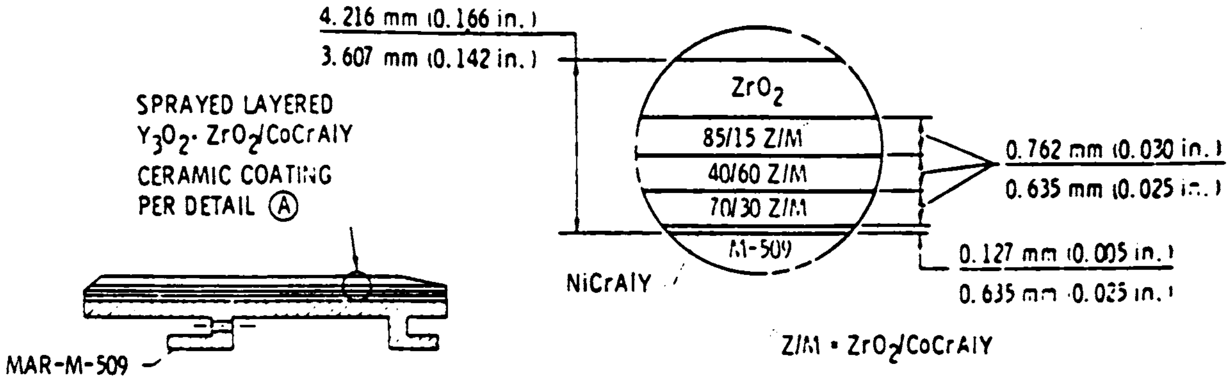


Figure 4-18. Typical engine seal segment.

An advanced ceramic-metal shroud is shown in Figure 4-18. It is produced by thermal-spraying ceramic-metal layers on a metal substrate of a heat-resistant metal alloy. The first step in the process is to spray a 0.127-mm (0.005-in.) coating of NiCrAlY on the metal substrate. This is followed by a layer composed of 60 percent CoCrAlY and 40 percent yttria-stabilized zirconia (ZrO_2). Next come successive layers of 30 percent CoCrAlY-70 percent ZrO_2 and 15 percent CoCrAlY-85 percent ZrO_2. Finally, the last layer, the one exposed directly to the turbine gas, is 100 percent ZrO_2 (Y_2O_3-stabilized).

This graded layer system provides a gradual change in thermal expansion coefficient and mitigates the large thermal expansion difference between the metal substrate and the ceramic layer next to the hot gas stream. Experimental studies show that graded layer ceramic material has adequate erosion resistance at 1315°C (2400°F) surface temperature. With the incorporation of controlled residual stress distribution through the ceramic seal system, adequate cyclic thermal shock resistance is realized for application to advanced commercial gas turbine engines. Alternate concepts involving different methods of reducing thermal stresses under more severe military engine cycles are under development.

- Conclusion--Gas path seals are necessary to the efficient operation of gas turbine engines. However, their development is still in an embryonic stage with respect to the long-life requirements of the large land-based gas turbines.

- Recommendation--The importance of gas path seals and design of small clearances dictate that a strong emphasis be placed on the development of suitable designs of seals, including ease of repair or replacement features for use in long-life, large land-based gas turbines.

LIFE PREDICTION AND TESTING

Analytical methods for predicting the life of critical high-temperature components of gas turbines are available and used but need to be extended for the development of the next generation of gas turbine designs (41). The critical components of concern are blades, vanes, disks (wheels), combustors, and transition sections. The new generation of turbines is expected to operate reliably between 100,000 and 200,000 hours with shutdowns and start-ups perhaps as often as 100 times per year and at turbine inlet temperatures up to 2600°F. The principal damage mechanisms are creep and thermal fatigue, complicated by environmental effects (especially oxidation and some unavoidable corrosion effects), brittle fracture, and long-time material structural instabilities as well as embrittlement reactions (e.g., temper embrittlement, aging).

Life prediction methods that are applicable to the anticipated requirements and operating conditions for the next generation of gas turbines need to be improved. This is especially true when considering the introduction of new classes of materials such as ceramics and composites. But even for the high-temperature materials currently used in aircraft gas turbines, the life prediction methods used by that industry may not be applicable to service periods longer by up to an order of magnitude or to the range of operating conditions characteristic of large land-based gas turbines. For example, 1000-hour creep data are inadequate for the prediction of creep behavior of a component with a required life range of 70,000 to 200,000 hours. It is important to note that, because of environmental effects and possible metastability of certain phases, long-time failure characteristics may be different from those measured in short-term laboratory tests.

The following topics need to be considered:

1. Stress analysis for the anticipated design and service conditions, including temperature cycles and potential excursions. The stress state (biaxiality and triaxiality) in critical regions is also of great concern.

2. Damage accumulation from creep and creep-fatigue interactions caused by normal as well as emergency-type operating conditions.

3. Crack growth behavior; thermomechanical cycles, environment, and short crack effects.

The materials of concern include not only those currently used in land-based and aircraft gas turbines but also composites, ceramics, and perhaps ordered alloys and materials made from rapidly solidified powders. Moreover, the effects of various coatings also need to be known. Test programs to determine the properties required for life estimates and to check out the reliability of the analytical methods developed will enable designers to obtain large pay-offs in the design of new land-based gas turbines. Attention must also be paid to the continued development of improved NDE methods for better reliability of life estimates and formulation of effective inspection and maintenance schedules.

Stress and Strain Analysis

A thorough stress-strain-temperature analysis for the anticipated operating conditions is required. Such analyses must include the time-temperature-stress histories for start-up, steady-state operations, shutdown, emergency trips, fast emergency start-ups, and foreseeable accidents. For critical locations, not only the stress (or strain) but also the stress state must be determined. For the thermoelastic regime, available finite element techniques are probably adequate to meet this requirement (42).

An understanding and appreciation of transient effects is of special importance. Thermal fatigue is one of the most frequent causes of failures of the high-temperature components (43). Temperatures as well as temperature gradients must be considered. The typical temperature distribution in a cooled vane is shown in Figure 4-19. The transient strain-temperature cycles are very sensitive to operating conditions. An unscheduled, sudden shutdown ("trip") exacerbates the strain-temperature history of a normal shutdown, as shown in Figure 4-20.

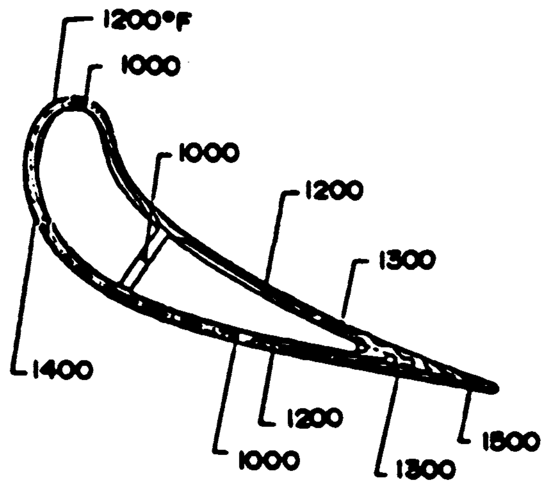


Figure 4-19. Cooled vane showing typical isotherms.

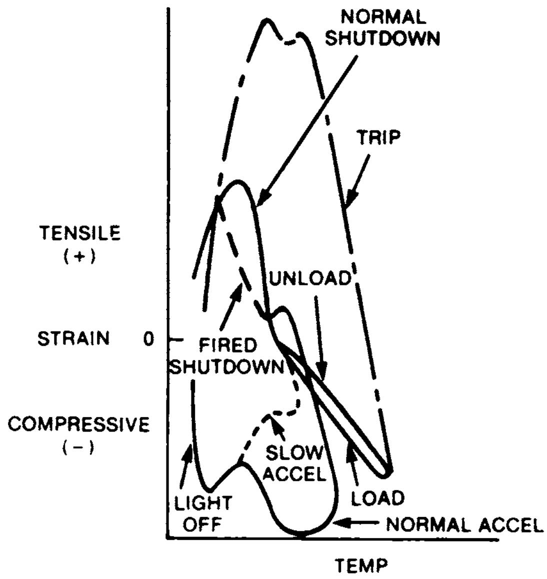


Figure 4-20. Leading-edge strain-temperature history, first-stage turbine blade.

Knowledge of accurate constitutive relationships in the inelastic regime is critical to account accurately for the time- and cycle-dependent stress-strain response in the overall design as well as for many design details--e.g., shrink fits and dovetails (blade-wheel attachment sections). At present there is concern with the accuracy of commonly used relationships, typically developed to represent sustained load tests, for more general load histories. In developing this area it is important to balance needs for the intended application with the costs involved in experimentation to obtain and validate accurate constitutive relationships and the complexity of stress analysis that results when they are applied.

Creep and Creep-Fatigue Interactions

Coffin et al. (44) prepared a general assessment of time-dependent fatigue of structural alloys. Mentioned among the unresolved issues at the time (1977) requiring more analytical work and testing were long-time testing; cumulative damage; multiaxial stress- and time-dependent fatigue; effects of environment; fatigue crack growth at elevated temperature; wave shape, unbalanced loops, and thermomechanical cycling; and low-strain high cycle fatigue. Although much work has since been accomplished, all of these topics are still of concern at present (41). Moreover, both creep and creep-fatigue as well as the nucleation and propagation of cracks must be included in a complete damage analysis. The latter are known to be particularly sensitive to environmental effects and are discussed in detail in a following section.

Analytical methods are usually based on summation of the incremental damage, i.e., fraction of failure strain (45). The principal damage components are the creep strain, which depends on the temperature and the stress, and the fatigue strain, which depends on the temperature gradient and the mechanically induced strain. For the former, the creep parameter approach has been used extensively to estimate long-term behavior. The parameters derived by various authors are listed as follows:

Dorn-Sherby (46):

$$\Theta = t \exp(-\Delta H_c/RT) = f_1(\delta)$$

Larson-Miller (47):

$$(T_F + 460)(C + \log t) = f_2(\delta)$$

Manson-Haferd (48):

$$\frac{T_F - T_a}{\log t - \log t_a} = f_3(\delta)$$

where

θ = temperature-compensated time parameter

t = time, hours

ΔH_C = activation energy for rate-controlling process

R = universal gas constant

T = absolute temperature

T_F = temperature, °F

C = a constant value between 10 and 30, usually 20

T_a, $\log t_a$ = constants derived from test data

Manson's "minimum commitment method" (49) represents an approach that unifies the above as well as other proposed parameters through a larger number of constants (up to six) that must be determined experimentally.

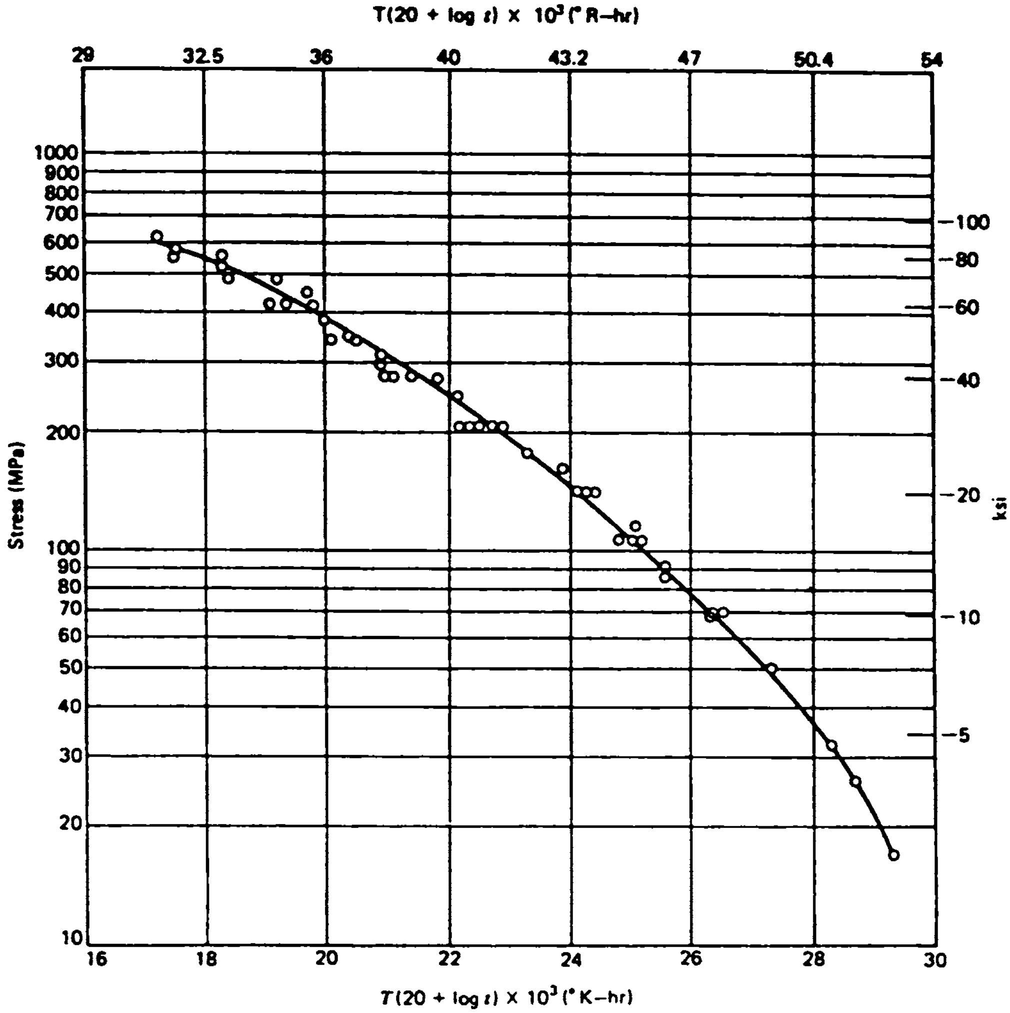


Figure 4-21. Larson-Miller plot, S-590 iron-based alloy (6).

Creep rupture data are presented in plots of one of these parameters, to integrate time and temperature, as a function of stress, as shown in Figure 4-21. For rupture under sustained stress the Monkman and Grant rule (50) is often applied:

$$\log t_r \quad + \quad \log (mcr) = \text{constant}$$

where t_r represents the time to rupture, (mcr) the minimum creep rate, and m and c are material-dependent constants. Since m is close to unity, $t_r \times (mcr) = \text{constant}$ is a fair approximation for many data. In some cases failure for intermittent creep may be estimated from

$$\Sigma e_i = e_{cr,f}$$

where e_i is the strain accumulated during an individual period and $e_{cr,f}$ is the creep failure strain; i.e., failure occurs when the sum of the strains accumulated during individual periods reaches the creep failure strain (51). However, the accuracy of such estimates is rarely adequate for long-term design purposes.

The anticipated deformation and failure regimes may be obtained from Ashby maps (52) wherever these are available for the materials and operating conditions of large land-based gas turbines. These maps, Figure 4-22, show that creep rate is suppressed by multiple strengthening mechanisms and grain coarsening.

For thermal and low cycle fatigue, the Coffin-Manson approach is generally used (53). For each cycle the strain range is separated into its elastic, ϵ_i, and plastic, ϵ_p, components. Fatigue life follows power laws of the type

$$N^{\alpha} \cdot \epsilon_p = \text{const.}$$

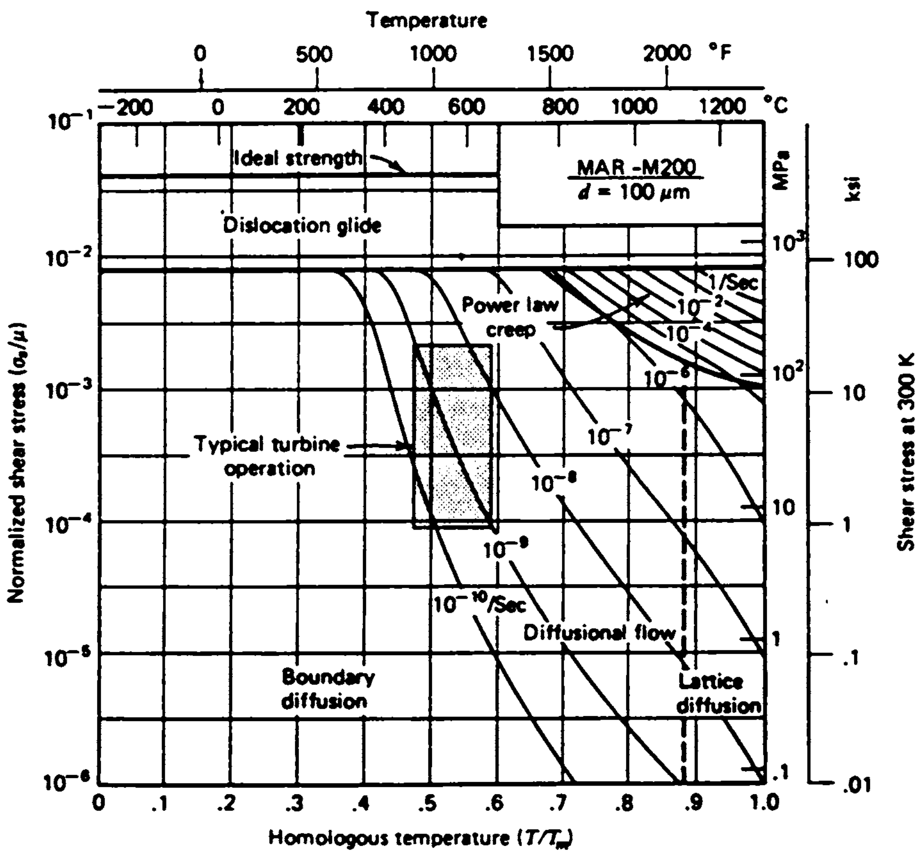


Figure 4-22. Deformation map.

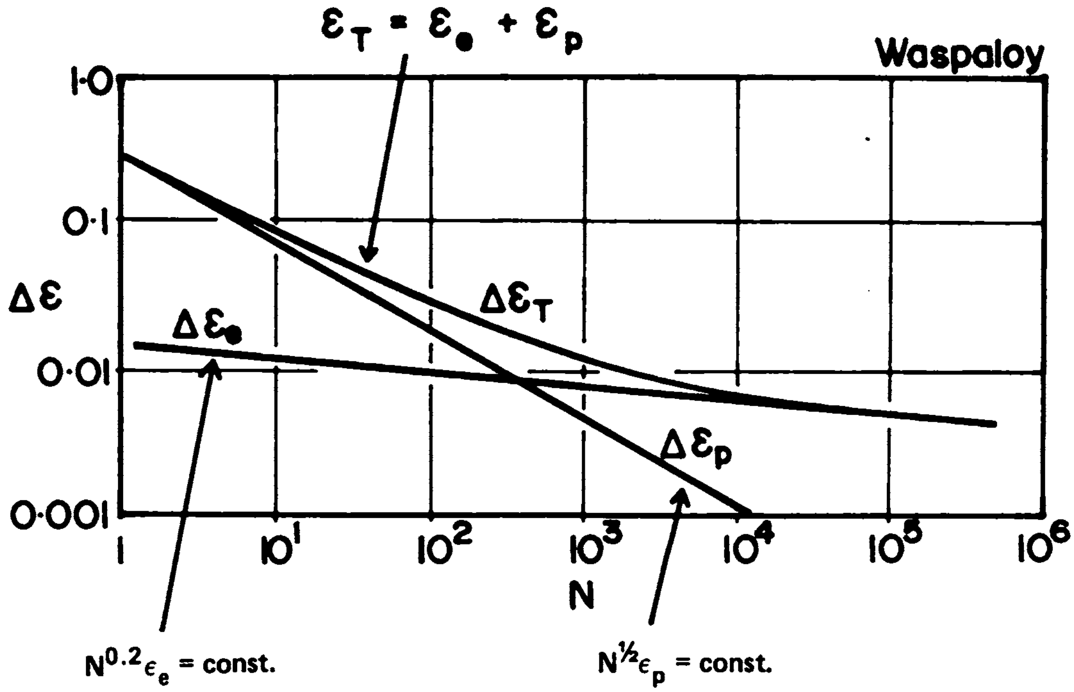


Figure 4-23. Relationship between $\Delta\epsilon$ and N.

and

$$N^{\beta} \cdot \epsilon_e = \text{const.}$$

where the exponents α and β depend on the material as well as on whether the strain range is elastic, ϵ_e, or plastic, ϵ_p (53). Figure 4-23 shows a typical representation of this strain range partitioning approach, where $\epsilon_t = \epsilon_e + \epsilon_p$; i.e., the total strain, ϵ_t, is the sum of its elastic, ϵ_e, and plastic, ϵ_p, components.

Crack Growth

A common mode of failure is by the nucleation and growth of cracks under the complicated thermal-mechanical loading and combustion gas environment conditions existing in turbine engines. The methodology for dealing with crack growth, using the elastic stress intensity factor to correlate a laboratory test specimen and service structure, is well established for moderately long cracks (say, greater than 0.5 mm) growing under nominally elastic conditions with rates in excess of 10^{-6} mm/cycle. Recent reviews that emphasize the creep-fatigue interaction in crack growth include those by Sadanada and Shahinian (54) and Saxena and Bassani (55); discussions of phenomena in specific alloys are given by Pelloux and Huang (56), Clavel et al. (57), and Floreen (58) in AIME volumes providing wider coverage of the area.

It is possible to test over the entire life for the comparatively short duty cycles of aircraft gas turbines. The much greater desired lifetime of large land-based turbines limits the feasibility of full-life testing and places a premium on accurate life prediction methodologies based on a quantitative understanding of crack growth. Further, the increased level of material inhomogeneity and segregation, together with greater inspection difficulties attendant to large component sizes, makes the presence of initial defects relatively more important in the land-based case. Lifetime may in many cases be determined by the growth to failure of initial defects rather than by the time to nucleate cracks in relatively defect-free material.

To improve crack-growth-based lifetime prediction it is necessary to have reliable stress analysis and NDE methodologies and also to develop a more robust quantitative framework for predicting crack growth. Some issues in crack growth that are not adequately understood at present involve the following:

1. Large-scale plastic deformation or extensive creep flow, which invalidates the linear elastic basis of the conventional methodology.

2. Procedures for extrapolating hold time or frequency and load wave from effects beyond the range of lab tests.

3. Threshold stress intensity factors, below which cracks do not grow (or grow less rapidly than about 10^{-10} mm/cycle), and their relation to environment.

4. The behavior of very short cracks (e.g., of the order 10 to 100 m length), which do not follow rules based on the stress intensity factor as the crack growth characterizing parameter.

Progress has been made on all four of these issues, but not necessarily in a form suitable for the turbine engine environment. For example, methods based on crack tip opening displacement and J integral have been advanced to deal with the non-elastic regime of item 1, and those methods have been extended to cyclic plastic loading at low temperatures (59,60) and, in the form of the C* integral, to sustained-load high-temperature crack growth (54).

With regard to item 2, a basis for rationalizing hold time and life effects at elevated temperature has been advanced (55) based on the two likely sources for these effects. The first source of time dependence involves material creep deformation in the concentrated stress field of the crack tip. This causes stress levels immediately after step loading to be greater than those prevailing later as relaxation proceeds and also causes the time-dependent development of strain. In analyses of these processes it is found that, following step loading, a "creep zone" develops at the crack tip and enlarges in size at a rate controlled by $\dot{K}$ and by creep deformation constitutive parameters. Within the creep zone, stresses are relaxed and large strains develop. It is important to extend such descriptions of the crack tip processes beyond simple step loading, to deal with actual complicated thermal and mechanical load histories.

The second source of time dependence involves the interaction with the environment, both in the form of surface interactions (presumably rapid) and diffusive transport of oxygen, and possibly other elements such as sulfur and sodium, into the material along grain boundaries (58,61). Fatigue crack growth along oxidized grain boundaries, supplied with the help of film ruptures at the cyclically loaded crack tip, is often observed at elevated temperature. The sensitivity of some fatigue crack growth data to environment (e.g., Inconel 718 at 650°C and Astroloy at 655°C, when tests in air are compared to those in helium or vacuum) suggests the importance of this second source of hold-time and frequency effects. On the other hand, for Cr-Mo-V steel at 538°C and for 304 stainless steel at 538°C and 570°C, fatigue crack growth appears to exhibit frequency or hold-time effects governed by creep-constitutive response of the material. The materials examples just cited are taken from Saxena and Bassani (55).

The issue of a fatigue threshold, item 3, is important for turbine applications since one is then often concerned with very slow growth of cracks and since the threshold seems to provide a basis for lack of effect of the many low-amplitude vibration cycles to which turbine engines are subjected. The threshold ΔK level decreases (sometimes to a plateau) with increasing load ratio R ($=K_{min}/K_{max}$) and with increasing strength levels. Effects of environment on the threshold ΔK have been paradoxical when viewed from the perspective of crack growth at higher rates, where chemically reactive environments (moisture, H_2 gas) should be regarded as enhancing growth. The opposite can happen near threshold. For example, the threshold ΔK for steel in a dry air environment at room temperature can be lower than for a moist H_2 environment. To our knowledge, analogous studies of near-threshold behavior have not yet been done for superalloys in combustion gas environments. However, a resolution of the room-temperature threshold results has been developed by Suresh et al. (62,63) and Ritchie (64). They show that the critical issue is the thickness of oxide layer that forms on the fatigue crack surface. This thickness is environment-sensitive and can be very much larger than the thickness on a surface not undergoing local plasticity and film break-up through surface contact. The greater the oxide film thickness, the greater the threshold ΔK. Evidently the oxide film, thicker than the preoxidized parent material layer, causes mechanical contact and crack closure. As is well established, fatigue crack growth is sensitive only to that part of the ΔK range for which the crack surfaces are actually separated near the tip.

It is considered important to have the results of similar studies for high-temperature turbine conditions. In particular, the role of oxide layers in affecting threshold ΔK levels and near-threshold growth needs to be determined.

With regard to item 4, the standard procedures of long-crack elastic fracture mechanics are now known to be inadequate for very short cracks. A review directed to room-temperature fatigue crack growth behavior has been given by Hudak (66) and a more detailed discussion of mechanisms by Suresh and Ritchie (60). The general effect is that the growth per cycle at a given K_{max} and K_{min} is larger for a very short crack (say, of the order 500μ m or less in length) than for a longer crack. In empirical terms such effects can to some extent be understood by basing crack growth on stress or deformation averaged over some distance ahead of the tip (a Neuber size) rather than on K. An approximately equivalent empirical procedure, which correlates short crack data in many cases (65,66) is to replace crack length a with a + L_0 in all elastic fracture mechanics formulas for predicting growth from long crack test data, where L_0 is a constant given as $(1/\pi)\ (\Delta K_{th}/\Delta\sigma_{th})^2$, and where ΔK_{th} is the threshold ΔK and $\Delta\sigma_{th}$ is the smooth-specimen endurance limit. However, it has been questioned (67) whether such procedures properly represent the physics involved, and a more fundamental understanding has not yet been attained.

In addressing all the above issues on crack growth, and hence lifetime prediction, it is important to incorporate the complicated thermal-mechanical loading histories expected for actual components in the turbine. Some fatigue testing has been done incorporating such histories (6,7,68).

Nondestructive Evaluation

It is important that advances in life prediction methodology, as discussed here, be coordinated with those in NDE for detection of defects so that a sounder basis is developed for interpreting inspections and for replacement or refurbishment decisions. A large amount of work on NDE is in progress throughout mechanical technology on such issues as inspection for defects (e.g., cracks or porosity), coating thickness measurements, and residual stress measurement in near-surface areas. Ultrasonic, magnetic, eddy current, X-ray, and penetrant techniques are common in this field.

Some possible advances applicable to the large land-based gas turbine operating regime involve detection of material aging or internal cavitation and of surface fatigue damage prior to identifiable cracking. For example, the degradation of strength properties with exposure time in superalloy land-based turbine blading might be estimated by monitoring the growth of precipitates by small-angle neutron scattering. Also, work is now in progress (V. Weiss, private communication) on using X-ray diffraction techniques on structural aluminum alloys and steels intended for aircraft structures to detect fatigue damage prior to crack nucleation. Feasibility studies of such approaches for high-temperature alloys in turbine operating conditions should be encouraged.

Other Issues

Oxidation-blocking coatings are widely used for high-temperature components, and some use is also made of thermal barrier coatings. It is important in all cases to better understand coating effects on crack nucleation and growth. For example, the beneficial effect of the coating in prohibiting internal oxidation may be compromised by cracking of the coating, amounting to earlier crack nucleation. For lifetime analysis, initial crack size is equated to coating thickness. Phenomena involved are discussed by Leverant et al. (6) and Sheirer and Moon (69). Thermal barrier coatings are also used to reduce component metal temperatures from those of the adjacent combustion gases. Here failure may involve separation of the coating due to oxidation of a bond layer between the coating and component (70,71).

The strong anisotropy of directionally solidified and monocrystal turbine blades will require an extension of stress analysis and fracture mechanics methodology to significantly anisotropic materials in ways that seem to have been as yet only partially addressed in aircraft turbine developments (7).

Future materials choices for turbines may involve fiber-reinforced metal matrix or ceramic composites and possibly monolithic ceramics based on SiN, SiC, etc. The problems posed for life prediction are serious for composites at room temperature and will presumably remain so at operating temperature. These involve complicated crack nucleation by fiber delamination, crack paths with steps of delamination along fiber directions, and incomplete fiber pull-out bridging a more complete matrix crack. Conventional fracture mechanics methodology seems to require significant extension to deal with such cases.

Furthermore, although it is less complicated than composites, the background for dealing with subcritical crack growth in monolithic ceramics remains somewhat less developed than for metallic alloys.

Conclusions and Recommendations Regarding Life Prediction and Testing

Improved life prediction methodology is important for the economical design and reliable operation of large land-based turbines as well as for the prevention of catastrophic fractures. This is especially important when material and design innovations are introduced. The following areas seem critical to advancing the methodology:

- Crack growth in the high-temperature range--Cases involving very small cracks and near-threshold conditions in the combustion gas environment are of interest. Time and thermal-mechanical waveform effects need to be related to the kinetics of environmental attack and material creep flow.

- Cumulative damage concepts--These approaches to predicting the nucleation of surface cracks must be extended to the particular environment-material combinations and thermal-mechanical waveforms anticipated in large turbine service. Crack initiation prediction needs to be combined with work on the growth of very small cracks to treat overall damage accumulation properly. It is also important to understand damage and crack nucleation in oxidation-blocking and thermal-barrier coatings.

- Constitutive relations--These are needed for accurate prediction of stress histories and strain accumulation of components in service, particularly in the most highly stressed locations. They are also expected to be part of the input for cumulative damage and crack growth prediction. Needs for accuracy have to be balanced with sound engineering judgment against the costs of extensive supporting experimentation and of resulting complexity in stress analysis.

- Long-term tests--These are needed in a form that can evaluate long-term environmental attack and materials microstructure evolution in conditions of strain and temperature variation appropriate to extended service conditions. The tests must be designed carefully to examine critical assumptions of the developing prediction methodologies for damage accumulation. They must also be designed to reveal mechanisms of cracking and environmental attack, including the occurrence of these processes on coated materials.

It is therefore recommended that experimental and analytical studies be carried out on key candidate materials for large land-based gas turbines to develop an understanding of crack growth and cumulative damage for the long-life operating conditions under consideration. The studies will specifically involve

- The development of more accurate constitutive relationships for candidate materials;

- Long-term tests of candidate materials; and
- Studies of environmental and long-term stability effects.

These studies should be limited to the most promising candidate materials. All work in this area, especially that conducted in connection with aircraft engine development, should be monitored.

REFERENCES

1. M. Gell, D. N. Duhl, and A. F. Giamei. "The Development of Single Crystal Superalloy Turbine Blades." Superalloys 1980, ASM, 1980, pp. 205-214.

2. F. L. VerSnyder. "Superalloy Technology--Today and Tomorrow." High Temperature Alloys for Gas Turbines 1982. D. Reidel Pub. Co., 1982, pp. 1-49.

3. F. L. VerSnyder and R. W. Guard. "Directional Grain Structures for High Temperature Strength." Trans. ASM, Vol. 52, 1960, p. 485.

4. M. Gell and G. R. Leverant. "The Fatigue of the Nickel-Base Superalloy Mar-200 in Single Crystal and Columnar-Grained Forms at Room Temperature." Trans. AIME, Vol. 242, 1968, p. 1869.

5. G. R. Leverant and M. Gell. "The Elevated Temperature Fatigue of a Nickel-Base Superalloy, Mar-M200, in Conventionally-Cast and Directionally-Solidified Forms." Trans. AIME, Vol. 245, 1969, pp. 1167-1173.

6. G. R. Leverant, T. E. Strangman, and B. S. Langer. "Parameters Controlling the Thermal Fatigue Properties of Conventionally-Cast and Directionally-Solidified Turbine Alloys." Superalloys: Metallurgy and Manufacture, Claitor's Pub. Div., 1976, pp. 285-295.

7. A. E. Gemma, B. S. Langer, and G. R. Leverant. "Thermomechanical Fatigue Crack Propagation in an Anisotropic (Directionally Solidified) Nickel-Base Superalloy." Thermal Fatigue of Materials and Components, ASTM STP 612, 1977.

8. K. A. Green, P. J. Loftus, and E. S. Nichols. "Development of Improved Marine Gas Turbine Airfoils Using Directional Solidification." Progress Reports on Naval Sea Systems Command Contract N00024-78-C-5636.

9. O. Y. Chen and D. N. Duhl. "Single Crystal Superalloy Turbine Airfoils for Marine Gas Turbine Engines." Progress Reports on Naval Sea Systems Command Contract N00024-78-C-5346.

10. J. S. Erickson, C. P. Sullivan, and F. L. VerSnyder. "Modern Processing Methods and Investment Casting of the Superalloy Family." High-Temperature Materials in Gas Turbines, Elsevier, 1974, pp. 315-343.

11. M. Cohen. Presentation at Workshop on Rapid Solidification Technology, National Bureau of Standards, July 2, 1981.

12. Powder Metallurgy Rene 95 Rotating Parts. NASA CR 159802, Vol. I, June 1979.

13. Rapid Solidification Processing: Status and Facilities. National Materials Advisory Board, NMAB-401, 1982.

14. J. S. Benjamin. "Dispersion Strengthened Superalloys by Mechanical Alloying." Met. Trans. 1, 1970.

15. J. K. Tien. "Alloy Design With Oxide Dispersoids and Precipitates." Fundamental Aspects of Structural Alloy Design, R. I. Jaffee and B. A. Wilcox, eds. Plenum Press, 1977, pp. 363-388.

16. T. G. Beatty and P. P. Millan. "Progress in the Utilization of an Oxide-Dispersion-Strengthened Alloy for Small Engine Turbine Blades." Aerospace Congress and Exposition, Long Beach, California, October, 1984, paper S41252.

17. Structural Uses for Ductile Ordered Alloys. National Materials Advisory Board, NMAB-419, 1984.

18. S. C. Harrier, General Electric Co. Private Commumication.

19. K. F. Mordoff. F404 Engine May Include Composite Material Duct. Aviation and Space Tech. Week, Sept. 17, 1984, p. 69-71.

20. R. W. Rice. "Capabilities and Design Issues for Emerging Tough Ceramics." Amer. Ceramic Bull., Vol. 63, No. 2, 1984, pp. 256-262.

21. R. W. Rice. "Mechanisms of Toughening in Ceramic Matrix Composites." Ceramic Eng. and Science Proceedings, Vol. 2, No. 7-8, 1981, pp. 661-701.

22. R. P. Ingel, D. Lewis, B. A. Bender, and R. W. Rice. "Temperature Dependence of Strength and Fracture Toughness of ZrO_2 Single Crystals." Journal of Amer. Ceramic Soc., Vol. 65, No. 9, 1982, pp. C150-C152.

23. D. W. Petrasek and R. A. Signorelli. Stress Rupture and Tensile Properties of Refractory Metal Wires at 2000° and 2200°F (1093° and 1204°C). NASA TN D-5139, 1969.

24. D. W. Petrasek. High Temperature Strength of Refractory Metal Wires and Consideration for Composite Applications. NASA TN D-6881, 1972.

25. D. W. Petrasek and R. A. Signorelli. Tungsten Fiber Reinforced Superalloys--A Status Review. NASA TM-82590, 1981.

26. R. A. Signorelli. Metal Matrix Composites for Aircraft Engines. NASA TM-83379, 1983.

27. E. A. Winsa, L. J. Westfall, and D. W. Petrasek. Predicted Inlet Gas Temperatures for Tungsten Fiber Reinforced Superalloy Turbine Blades. NASA TM-73842, 1978.

28. P. Melnyk and J. N. Fleck. Tungsten Wire/FeCrAlY Matrix Turbine Blade Fabrication Study. NASA CR-59788, 1979.

29. D. W. Petrasek, E. A. Winsa, L. J. Westfall and R. A. Signorelli. Tungsten Fiber Reinforced FeCrAlY--A First Generation Composite Turbine Blade Material. NASA TM-79094, 1974.

30. D. W. Petrasek and J. W. Weeton. Alloying Effect on Tensile Properties and Microstructure of Tungsten-Fiber-Reinforced Composites. NASA TN D-1568, 1963.

31. L. J. Westfall. Tungsten Fiber Reinforced Superalloy Composite Monolayer Fabrication by an Arc-Spray Process. NASA TM-86917, 1985.

32. F. S. Pettit and G. W. Goward. "High Temperature Corrosion and Use Coatings for Protection. Met. Treatises, J. K. Tien and J. F. Elliott eds. The Metallurgical Society, Warrendale, Pennsylvania, 1982, pp. 603-621.

33. E. Lang, ed., Elsevier Press, New York, 1982.

34. C. Sims and R. Hagel. The Superalloys, Chap. 12. John Wiley & Son, New York, 1972.

35. Research Directed to Advanced High Temperature Coating Systems Beyond Current State-of-the-Art Systems. Fourth Annual Report, AFOSR, Bolling AFB, D.C., October 1984.

36. Metallic Coating Development and Evaluation. Performed by Pratt & Whitney Aircraft for Naval Sea Systems Command, P.O. Box 2691, West Palm Beach Florida, April, 1984.

37. D. K. Gupta and R. S. Mullen. Manufacturing Technology for Vacuum Plasma Spray Overlay Coatings. Mfg.-Tech. Div., Materials Lab, ACSC, Wright-Patterson AFB, Ohio, April 1984.

38. D. K. Gupta and D. S. Duval. "A Silicon and Hafnium Modified Plasma Sprayed MCrAlY Coating for Single Crystal Superalloys." Superalloys 1984, M. Gill, C. S. Kortorich, R. H. Bricknell, W. B. Kent, and J. F. Radavich, eds. The Metallurgical Society of AIME, Warrendale, Pennsylvania, 1984, pp 713-722.

39. G. Lehnert and H. Meershardt. "A New Protective Coating for Nickel Alloys." Electrodeposition and Surface Treatment, Vol. 1, 1972-73, pp. 189-197.

40. T. Strangman. "Thermal Fatigue of Oxidation Resistant Overlay Coatings for Superalloys." Doctoral Dissertation, University of Connecticut, 1978.

41. Analysis of Life Prediction Methods for Time-Dependent Fatigue Crack Initiation in Nickel-Base Superalloys. National Materials Advisory Board, NMAB-347, 1980.

42. Users Manual, Example Problems Manual, Systems Manual, and Theory Manual for the ABAQUS General Purpose Nonlinear Finite Element Program. Hibbit, Karlsson & Sorensen, Inc., Providence, Rhode Island.

43. D. E. Brandt. Gas Turbine Design Philosophy. General Electric Gas Turbine Reference Library, Report GER-3434, 1984.

44. L. F. Coffin, Jr., S. S. Manson, A. E. Carden, L. K. Severud, and W. L. Greenstreet. Time-Dependent Fatigue of Structural Alloys: A General Assessment (1975). Oak Ridge National Laboratory, Report No. ORNL-5073, January 1977.

45. Class 1 Components in Elevated Temperature Service, Class III, ASME Boilers and Pressure Vessel Code, Case Interpretations, Code Case 1592. American Society of Mechanical Engineers, New York.

46. R. L. Orr, O. D. Sherby, and J. E. Dorn. "Correlations of Rupture Data for Metals at Elevated Temperatures." Trans. ASM Vol. 46, 1954, pp. 113-128

47. F. R. Larson and J. Miller. Trans. ASME, Vol. 74, 1952, pp. 765-771.

48. S. S. Manson and A. M. Haferd. NACA Tech Note 2890, March 1953.

49. S. S. Manson and C. R. Ensign. "A Quarter Century of Progress in the Development of Correlation and Extrapolation Methods for Creep Rupture Data." Journal of Engineering Materials and Technology, Vol. 101, October 1979, p. 317.

50. J. C. Monkman and N. J. Grant. Proc. Am. Soc. Testing Materials, Vol. 56, 1956, p. 593.

51. G. Sachs and V. Weiss. "Beiträge zur Kurzzeitermüdung." Zeitschrift für Metallkunde, Vol. 53, No 1, 1961, p. 1307.

52. M. F. Ashby and B. Tomkins. "Micromechanisms of Fracture and Elevated Temperature Fracture Mechanics." Proc. Third Int. Conf. on Mechanical Behavior of Materials, Vol. 1, Cambridge, England, 1979, pp. 47-89.

53. "Fatigue: An Interdisciplinary Approach." In Proceedings of the Tenth Sagamore Army Materials Research Conference, J. J. Burke, N. L. Reed, and V. Weiss, eds. Syracuse University Press, 1964.

54. K. Sadanada and P. Shahinian. "Creep-Fatigue Crack Growth." Cavities and Cracks in Creep and Fatigue, J. Gittus, ed. Applied Science Publishers, London, 1981, pp. 109-195.

55. A. Saxena and J. L. Bassani. "Time Dependent Fatigue Crack Growth Behavior at Elevated Temperature." To appear in Synergism of Microstructure, Mechanisms and Mechanics of Fracture, J. M. Wells, ed. AIME, preprint 1984.

56. R. M. Pelloux and J. S. Huang. "Creep-Fatigue-Environment Interactions in Astroloy." Creep-Fatigue-Environment Interactions, R. M. Pelloux and N. S. Stoloff, eds. Met. Soc. AIME, 1980, pp. 151-164.

57. M. Clavel, C. Levaillant, and A. Pineau. "Influence of Micromechanics of Cyclic Deformation at Elevated Temperature on Fatigue Behavior." Creep-Fatigue-Environment Interactions, R. M. Pelloux and N. S. Stoloff, eds. Met. Soc AIME, 1980, pp. 24-45.

58. S. Floreen. "Effects of Environment on Intermediate Temperature Crack Growth in Superalloys." Micro and Macro Mechanics of Crack Growth, K. Sadanada, B. B. Rath, and D. J. Nickel, eds. Met. Soc. AIME, 1981, pp. 177-184.

59. N. E. Dowling. "Geometry Effects and the J-Integral Approach to Elastic-Plastic Fatigue Crack Growth." Cracks and Fracture, ASTM-STP-601, 1976, pp. 19-32.

60. S. Suresh and R. O. Ritchie. "Propagation of Short Fatigue Cracks." International Metals Reviews, Vol. 29, No. 6, 1984, pp. 445-476.

61. C. J. McMahon. "On the Mechanism of Premature In-Service Failure of Nickel-Base Superalloy Gas Turbine Blades." Mater. Sci. Engr., Vol. 13, 1974, p. 295.

62. S. Suresh, G. F. Zamiski, and R. O. Ritchie. "Oxide-induced Crack Closure: An Explanation for Near-Threshold Corrosion Fatigue Crack Growth Behavior." Metallurgical Transactions, Vol. 12A, 1981, p. 1435.

63. S. Suresh, D. M. Parks, and R. O. Ritchie. "Crack Tip Oxide Formation and Its Influence on Fatigue Thresholds." Fatigue Thresholds, J. Backlund, A. F. Blom, and C. J. Beevers, eds. Engineering Materials Advisory Service Publ. Ltd., Warley, U. K., 1982, Vol. 1, p. 391.

64. R. O. Ritchie. "Environmental Effects of Near Threshold Fatigue Crack Propagation in Steels: A Reassessment." Fatigue Thresholds, J. Backlund, A. F. Blom, and C. J. Beevers, eds. Engineering Materials Advisory Services Publ. Ltd., Warley, U.K., 1982, Vol. 1.

65. M. H. El Haddad, K. N. Smith, and T. H. Topper. "Fatigue Crack Propagation of Short Cracks." Trans. ASME, J. Eng. Mat. and Tech., Vol. 101, 1979, pp. 42-46.

66. S. J. Hudak. "Small Crack Behavior and the Prediction of Fatigue Life." Trans. ASME, J. Eng. Mat. and Tech., Vol. 103, 1981, pp. 26-35.

67. J. Lankford, D. L. Davidson, and K. S. Chan. "The Influence of Crack Tip Plasticity in the Growth of Small Fatigue Cracks." Met. Trans., Vol. 15A, 1984, pp. 1579-1588.

68. G. T. Embley and E. S. Russell. "Thermal-Mechanical Fatigue of Gas Turbine Bucket Alloys." First Parsons International Turbine Conference (Dublin, June 1984). Institution of Mechanical Engineers, London, 1984, pp. 157-164.

69. S. T. Scheirer and D. M. Moon. "Materials for Gas Turbines." Preprint for inclusion in Sawyer's Gas Turbine Handbook, 1984.

70. R. A. Miller and C. E. Lowell. "Failure Mechanisms of Thermal Barrier Coatings Exposed to Elevated Temperature." Thin Solid Films, Vol. 95, 1982, p. 265.

71. R. A. Miller. "Oxidation Based Model for Thermal Barrier Coating Life." J. Amer. Ceramic Soc., Vol. 67, 1984, pp. 517-521.

72. S. S. Manson. "A Complex Subject--Some Simple Approximations." Experimental Mechanics, No. 7, 1965, pp. 193-226.

Appendix

BIOGRAPHICAL SKETCHES OF COMMITTEE MEMBERS

Louis R. McCreight received his B.S. and M.S. degrees in ceramic engineering from the University of Illinois at Urbana-Champaign. After serving on the Manhattan Project and at the University of Illinois, he worked at the Knolls Atomic Power Laboratory and in the Missiles and Space Division of General Electric Company as well as with the Materials Sciences Laboratory at the Aerospace Corporation. His research interests include refractory materials, nuclear fuel elements, vehicle materials, and space processing.

Donald J. Jordan earned his B.S. degree at New York University in engineering. He served as a power plant engineer for Chance Vought Aircraft and on the staff of Pratt and Whitney before becoming engineering manager of the Power Systems Division of United Technologies. He is currently an adjunct professor at M.I.T. He is a member of the National Academy of Engineering.

Thomas F. Kearns received his A.B. and B.S. degrees from Columbia University. He worked as a metallurgist for the Crucible Steel Company and the Ford Instrument Company, and retired as Technology Administrator, Ferrodynamics Structures and Materials, Naval Air Systems Command, before joining the research staff of the Institute for Defense Analyses.

Gerald R. Leverant earned his B.M.E. and Ph.D. degrees in metallurgy at Rensselaer Polytechnic Institute. He worked at the United Aircraft Research Laboratories and at Pratt and Whitney Aircraft before joining Southwest Research Institute, where he is currently the director of the Department of Materials Science.

Francis D. Lordi received his B.S. and A.B. degrees from Columbia University. He has been employed at the General Electric Company's Foundry Department, Knolls Atomic Power Laboratory, and the Large Steam Turbine and Generator Division's Materials and Processes Laboratory. Currently he is manager of materials engineering in GE's Gas Turbine Division.

Frederick S. Pettit earned his B.E., M.Eng., and D.Eng. (metallurgy) degrees at Yale University. He worked at Westinghouse Atomic Power Division, Lycoming Division of Avco Manufacturing Corporation, United Aircraft Corporation, and Pratt and Whitney Aircraft, as well as in academic positions at Yale and at the Max Planck Institute in Göttingen, Germany. He is currently chairman of the Metallurgical and Materials Engineering Department at the University of Pittsburgh. His research interests are in the oxidation of metals and alloys and the thermodynamics and kinetics of solid-state reactions.

James R. Rice earned his Sc.D. degree in mechanical engineering at Lehigh University and did research and teaching at Brown University before joining Harvard University, where he is now professor of engineering science and geophysics. His research interests are in crack growth and fracture processes in solids, metal plasticity, and earthquake source mechanics. He is a member of the National Academy of Sciences and the National Academy of Engineering.

Roy W. Rice received his B.S. and M.S. degrees at the University of Washington. He worked at the Boeing Company before joining the Naval Research Laboratory, where he was head of the Ceramics Branch. He is now with the W. R. Grace Company. His research interests are in the relationships between processing microstructure and mechanical properties of ceramics.

Scott T. Scheirer earned B.S. and M.S. degrees at Lehigh University and his Ph.D. degree in mechanical engineering at Case Western Reserve University. He worked at TRW, Inc., before joining the Westinghouse Electric Corporation, where he is now manager of metallurgical engineering in the Combustion Turbine Systems Division. His research interests are in development and selection of materials for gas turbine engines.

Volker Weiss received his diploma in metallurgy at Vienna's Technological University and his M.S. and Ph.D. degrees at Syracuse University. He worked in private industry in Europe and the United States before joining the faculty of Syracuse University, where he is now vice president for research and graduate affairs. His research has been in metal physics, fracture mechanics, fatigue, residual stresses, solid-state reactions, superplasticity, and X-ray diffraction.