Materials Technology for Advanced Land Based Gas Turbines

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Explosive growth has occurred in the combustion turbine industry, especially for electric power generating plants. Many of these units are intended for base load operation in combined cycle configurations, and as a result have a premium on efficiency, reliability, and extended maintenance cycles. These targets are met through the use of advanced designs, materials, and coatings for hot section components. This paper reviews the evolution of these technologies along with a user perspective on reliability and maintenance issues.

1. INTRODUCTION

In recent years, gas turbines (GT) have become the equipment of choice for power generation by both electric utilities and independent power producers. Some interesting statistics reported in 1998 by Zink are noteworthy(1). According to his article, 11% of the total US power generation at that time was in the hands of independent power producers who typically opt for GT power plants. In the intervening years this percentage has increased. By the end of 2001 the Electric Power supply Association expects about 25% of the US installed generating capacity to be owned and operated as competitive merchant plants(1a). It is also anticipated that over the next 10 years, utilities will shut down nearly 25 GW of nuclear capacity reaching its end of life and that GTs will replace most of this. Worldwide, GTs now comprise 40% of all new capacity additions. This is twice the percentage contribution of GTs to total capacity 20 years ago, as shown in Figure 1.

The reasons for the popularity of GTs are not difficult to understand. Among these are: availability of clean burning gas fuel, short lead times and ease of installation of GT plants, modular nature of plant and flexibility to increase plant size incrementally in the face of uncertain load growth, lower CO_2 emissions and cost of environmental compliance, substantial fuel cost savings due to increased efficiency in combined cycle configuration, and availability of GT plants at power ratings as high as 280 MW. For base load applications, combined cycle plants approaching 60% thermal efficiency can be installed at a cost as low as $400 per kW(1). The rapid cycling ability of simple cycle GTs also makes them an attractive part of the generation mix.

Table 1 shows the chronological evolution of large land based GTs. Models that correspond approximately to these conditions are also indicated although continuous evolutions and upgrades make precise correspondence impossible. During the last 30 years the power rating of GTs have increased from 50 to nearly 280 MW and the thermal efficiencies have increased from 29 to 39% in simple cycle and from 43 to 58% in combined cycle configurations. These changes have been largely possible because of increases in the rotor inlet temperature from 900 to 1425°C (1650 to 2600°F), the pressure ratio from 10.5 to 23, and also air flow (not shown in Table 1). In models currently under development (e.g. MHI H series (1b) and Siemens-Westinghouse ATS), power will exceed 400 MW with efficiencies cracking the 60% barrier.

The current generations of land based gas turbines introduce advancements in cooling technology, aerodynamic design, and mechanical innovation. Materials technology has necessarily kept pace, with some evolutionary improvements derived from recent versions of the prior generation and some significant introductions of advanced concepts. As is the case with other aspects of gas turbine technology, some of these new materials and processes have been adapted from the aircraft engine industry, although rarely has there been a direct and painless transfer.

Specific materials advancements adapted from aircraft engine technology include: increased use of directionally solidified alloys for larger sized blades, single crystal blades, wider use of ceramic thermal barrier coatings, further improvements in coatings for oxidation and corrosion protection, and nickel based alloys for discs. Other innovations on tap for future land based gas turbines include: limited use of structural ceramics, increased use of advanced welding and joining methods, and condition assessment methods based on those used for aero-engines. The present paper will review the evolution of materials and coatings.

An additional factor to be reckoned with is the industry-wide consolidation that has taken place in recent years. Siemens has acquired the Westinghouse Power Generation operations with the result that Siemens V94 class turbines are being sold for the 50 Hz market and Westinghouse 501 designs are featured for the 60 Hz market. One would naturally expect an interchange of materials technologies. MHI is now in the market as a separate entity. Their materials choices are expected to differ from the S-W 501, but remain similar for the near term. As previously, General Electric maintains primary control over materials selections in GTs for both the 50 Hz and 60 Hz markets. Meanwhile, Alstom has acquired the gas turbine operations of ABB.

2. HOT SECTION MATERIALS

The key components exposed to the hot gas path are the combustor, transition pieces, turbine vanes, blades and disks. The materials technology relating to these components is discussed in this section.

2.1 Combustors

There are three basic designs for combustors—can type, annular and external silo type. Details of these may be found elsewhere (2). Current models use mainly can type or annular combustors and the materials issues involved are the same for both.

Combustor wall temperatures can be very high. The materials undergo very abrupt temperature changes on start and stop, so that low cycle fatigue (LCF) is an important failure mode. Steady-state thermal stresses can also be significant due to the nature of the combustion process and the need for wall cooling. A design that is flexible but still able to withstand thermal fatigue is a must. The combustion process itself generates high-frequency vibrations, which can result in high cycle fatigue (HCF) failure. The relatively thin walls of the combustor can and liner mean that oxidation is also an important failure mode. Finally, the pressure outside the combustor is somewhat higher than that inside. Since the walls are thin, this pressure difference is sufficient for creep rupture and buckling to be a concern, where temperatures are highest.

For GE engines the materials that have been used in the construction of combustors are the nickel-based alloys Hastelloy X, RA333 and Nimonic 263. [1] For the MS7001F and MS9001F

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[1] Chemical compositions of all the hot-section component alloys described here may be found in References 3 and 4. Reference 4 also describes damage mechanisms in detail.

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engines the cobalt base alloy HS-188 is used in the latter section of the liner for improved creep-rupture strength(5). In the MS7001H and MS9001H, Olsen (5a) reports that the combustion liners and transition pieces will be air cooled cast GTD-222. Pallotta (6) reported that Westinghouse (S-W) engines primarily use Hastelloy X in their combustors. MHI uses the nickel-base alloy Tomilloy, a derivative of IN617 for combustors, which are steam cooled in the 501/701G and the 501/701H which is under development (1a). Thermal barrier coatings are widely used to maximize the effect of external cooling air, maintaining metal temperatures within creep limits and helping to mitigate thermal gradients. These are usually around 0.010 in. of plasma sprayed $ZrO_2 \bullet Y_2O_3$, with a bond coat of plasma sprayed NiCoCrAlY. The annular combustor designs now featured in Alstom and Siemens turbines use similar types of nickel base alloys for the outer shell but, borrowing from the external combustor designs, line them with refractory ceramics. Cooling air flows in the space between the inner and outer linings and is then used in the combustion process. A summary of the materials used in combustors in the current model GTs is provided in Table 2.

2.2 Transition Pieces

The combustion gas from the combustor exit is directed toward the first ring of stationary airfoils (nozzle guide vanes). It is important that the flow of hot gas into this stage is distributed as uniformly as possible. The ducts that perform this function are the transition pieces. In the case of an engine with can combustors, the transition pieces have a circular cross-section at one end, which joins the can. The other end has an approximately rectangular cross-section, which on its narrow sides adjoins the neighboring transition pieces, forming a more-or-less continuous ring facing the inlet nozzle guide vanes. The gas entering the transition piece is at the burner outlet temperature and leaves the transition at about the same temperature.

Although the transition pieces are less complicated than the liners, they have been more challenging from a materials/process standpoint because of the combination of stresses and temperatures encountered in service, and because less cooling is possible(5). For GE and Westinghouse type engines, Hastelloy X became standard material by 1970. In the early 1980s, GE changed to Nimonic 263, which is now being used in many high-firing temperature machines. HS-188 is also used in a few high temperature models, with cast GTD-222 planned for the H series. Around 1980 Westinghouse began using IN617 and continues to use this material and Haynes 230. ABB has also used IN617 for ducting in its external combustor models. In the late 1980's MHI began to use Tomilloy, a proprietary derivative of In617.

Improved cooling, thermal barrier coatings, and structural reinforcements have permitted these materials to remain useful. The transition cooling schemes for the "F" series turbines have become quite complex, involving impingement plates (GE) or internal channels for cooling air known as MTFIN (Westinghouse & Mitsubishi). This is a bonded and welded structure similar in some ways to Allison's Lamilloy of around 20 years ago. For the "G" class turbine, S-W and MHI use similar materials and construction, but employ externally generated steam as the cooling medium(7). Table 2 summarizes the materials used.

2.3 Stationary Vanes

The function of the stationary vanes is to take the hot gas from the combustor (or previous stage) and turn it (generally accelerating the gas as well) so that it reaches the next rotating stage at the optimum angle. The first stage vanes experience the highest gas temperature in the turbine itself, and the highest gas velocity. Since the vanes are stationary, the only stresses are due to the aerodynamic loading and thermal gradients. While these are not as high as the stresses in the rotating stages, they are by no means trivial. Schilke et al (5) list the properties required by the first stage vanes (called "nozzle guide vanes" by GE) as: 1) excellent oxidation and corrosion resistance, 2) high resistance to thermal fatigue, 3) relatively good weldability for ease of manufacture and repair, and 4) good castability. To this list one should add creep resistance to withstand steady state thermal stresses and the weight of the vane assemblies themselves which are cantilevered from the outer diameter.

GE's practice has been to use cobalt-base alloys for the stator vanes. X-40 and X-45 were used in the early years, but were eventually replaced by FSX-414, which has less carbon to improve weldability, and more chromium to enhance oxidation resistance. Currently, first stage nozzles through the "F" series and the majority of later stage nozzles are made of this alloy. Recently, a new nickel-base alloy, GTD-222 has been developed to provide improved creep strength in the second and third stage nozzles, and is currently being used in the later stages of the MS7F/9F machines. A new cobalt-base alloy with better properties than FSX-414 is being developed, named GTD-484: but this does not appear to have entered service yet(5).

The vane alloys used by S-W are ECY768 (a modified Mar-M 509 alloy), WES 100, and X-45(6). The choice of alloy and degree of cooling depend on mechanical design and performance considerations. In the W501G, all stages of vanes are cast nickel-base alloy, IN 939(7).

Holmes, Pfeffer and Schneider have summarized ABB's (Alstom) practice with regard to blading materials(8). They remark that for the first and second stage vanes the reference creep stress condition is 100 Mpa (14.3 ksi) at a temperature of 950°C (1742°F). Current practice is to use the cast nickel-base alloys IN 738 and IN 939. Casting porosity becomes more of a problem as the size of the castings increases, and the porosity can act as a site for fatigue crack initiation. This has led to the general introduction by ABB of Hot Isostatic Pressing (HIPing) into the production process, to close up the pores. This results in a very significant improvement in fatigue properties. The IN 738 LC and IN 939 investment cast vanes, are cooled to achieve average metal temperatures not exceeding 850°C (1562°F). Siemens uses the same alloys for the latest models of its large industrial gas turbines in the "Vx4 series"(9).

In the 1425°C (2600°F) H technology engines developed for combined cycle plants, GE is planning to introduce a single crystal N5 material for the stage 1 nozzle. This represents a significant advancement in manufacturing technology. The second stage nozzle is also noteworthy in that it consists of a DS GTD-222 airfoil joined to separately cast shrouds using advanced welding and brazing methods. Both components are steam cooled to achieve satisfactory metal temperatures(5a, 10). Westinghouse relies on advanced air cooling with showerhead leading edge design and thermal barrier coatings(7). In current models of the F and H series, MHI uses conventionally cast MGA2400 for all for stages of stationary vanes.(1b) This material is a nickel-base alloy, believed to be similar to IN939. For the GT24/26 Alstom uses cast MarM247 LC for the first three stationary stages(11).

2.4 Rotating Blades

The rotating blades are more highly stressed than the vanes, but experience a slightly lower gas temperature. Required properties include those listed for the stationary vanes (oxidation and corrosion resistance, thermal and low cycle fatigue resistance, repairability, ease of manufacture-castability, and creep resistance) and also high cycle fatigue resistance to withstand various sorts of vibratory loading. Furthermore, in today's highly cooled airfoils, there is a need to tolerate large, highly localized stresses adjacent to cooling passages. Finally, the metallurgical properties of long term microstructural stability and coatability must be considered.

An extensive listing of vane and blade materials used in current designs is provided in Table 3. This list is tentative since evolutionary changes historically have led to periodic upratings and changes of material.

Figure 2 illustrates the increase in firing temperature over the years and the corresponding developments in blade (bucket) alloys (5). It can be seen from this figure that the increase in alloy high-temperature strength, which accounted for the majority of firing-temperature increases until about 1970, slowed during the 1970s. This occurred as the result of two factors. First, emphasis was placed on the use of air cooling to increase firing temperatures. Second, as the metal temperatures approached 870°C (1600°F), hot corrosion of buckets became a more life-limiting factor. Together with the increased use of more contaminated fuels, this required that material developments be directed more toward improvement of hot corrosion resistance and the use of long-life corrosion-resistant coatings. The 1980s experienced a swing in emphasis back toward development of stronger alloys in an effort to fill part of the need for machine uprating. Directionally solidified airfoils were introduced in GTs in 1987 as a replacement row 1 bucket in the MS5002. More recently single crystal superalloys have become a reality.

The evolution of conventional superalloys has been dictated by balancing requirements of strength, hot corrosion resistance, oxidation resistance, freedom from deleterious sigma phase and forgeability. Blades in the early engines were wrought nickel base alloys, related to Nichrome Ni-20Cr. It was discovered that these materials could be strengthened by coherent precipitation of $Ni_3(Al, Ti)$, known as γ'. Increasing the volume fraction of this phase required the reduction of chromium and addition of cobalt. The loss of Cr resulted in both a loss of solid solution strength and oxidation resistance which were compensated for by addition of molybdenum and aluminum respectively. The discovery of another form of corrosion called hot corrosion changed the picture once again. Two forms of hot corrosion, a type I occurring at 800-925°C (1472-1700°F) and a low temperature form of hot corrosion (type II) occurring at 700-775°C (1290-1425°F) required a higher chromium to aluminum ratio and substitution of molybdenum by other refractory elements W, Ta and Nb. Since forging the high strength alloys became increasingly difficult, castable alloys such as Inco 738, Inco 939 and Inco 792 were developed and utilized for many years. Further increases in turbine inlet temperatures have led to the development of directionally solidified (DS) and single crystal alloys.

In combustion turbines, the creep-rupture lifetime of both the wrought and cast blades is generally 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 (12). 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. Significant improvements with respect to resistance to thermal fatigue, crack growth and oxidation compared to conventionally cast alloys have also been noted. Table 4 lists the composition of some candidate alloys under investigation for both DS and single crystal applications(13). DS alloys initially resembled the composition of more advanced equiaxed materials. It was soon discovered, however, that the use of DS technique plus the addition of hafnium made it possible to use high strength eutectic compositions, producing significant increases in rupture strength. Lowering the Ti/Al and addition of Ta also helped optimize the alloys in terms of oxidation resistance.

DS first stage buckets are currently being used in the MS7001F and MS9001F and uprated MS3002 and MS5002 models offered by GE. It has been claimed that compared with its equiaxed counterpart, the alloy DS GTD 111 bucket has improved creep life, impact strength (by more than 33%) and thermal fatigue resistance (by more than 10 times). DS alloys also appear in the most recent designs from Alstom, Siemens-Westinghouse, and Mitsubishi.

Single crystal alloys offer further flexibility in alloy design and associated increase in temperature capability. The move from DS alloys to single crystals is promoted for the following reasons:

- The directional solidification process results in large MC-type carbides in some alloys, and these carbides are often precracked and initiate matrix fatigue cracks under cyclic loading conditions(14,15).
- If some grain boundaries are not perfectly aligned with the solidification direction, creep cracks may initiate at these boundaries where they intersect a free surface.
- 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. The increase in the melting point permits solutioning of large amounts of coarse γ' which later precipitate as fine γ' enabling realization of the full strength potential.
- 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. Development of leaner alloys also improves castability.

Clearly, alloy designers have taken advantage of these features as may be seen by comparisons of the compositions of the single crystal alloys in Table 5 (16) with those of the DS alloys in Table 4. Addition of rhenium has also proved a boon in improving creep strength realizable from single crystal alloys over DS and conventionally cast materials is illustrated in Figure 3. The alloy Nasair 100, which is a single crystal version derivative of Mar M 247 shows a 22°C (40°F) increase of temperature capability with respect to the conventionally cast Mar M 247 and about 17°C (30°F) with respect to DS Mar M 247.

Single crystal blades were initially put into commercial service in 1982 in Pratt & Whitney's engine for the Boeing 767 aircraft and the Airbus A 310 (alloy PWA 1480). Howmet, a leading producer of SC casting for aircraft components has shipped more than 10,000 single crystal castings. All major aircraft engine manufacturers have developed and applied new SC alloys in recent years. The only use of SC alloys in land-based turbines until recently has been the use of CMSX 4 alloy in Solar industrial turbine Mars T-14000 engine in 1990(17). Single crystal alloy blades made of PWA1483 are currently in use in Siemens V84/94.3a GTs. Alloy Rene N5 has found a place in GE 7H turbines as first row blading. Pallotta has reported on a Westinghouse development of a SC alloys and has shown properties but without the composition. At the time the Westinghouse DS alloy was reported to have a temperature capability of more than 50% greater than Inco 738LC and the SC alloy has a capability 100% greater(6). To this point it appears to remain developmental.

The benefits of SC have also been shown by Reference 18 (see Figure 4). The creep strength as well as the thermal fatigue resistance of the SC alloy show a nine-fold improvement with respect to the polycrystalline material. The oxidation/hot corrosion resistance is increased by a factor of three.

Figure 5 shows the conflicting trends in creep strength and hot corrosion resistance (based on Ref. 18). As the Cr content increases, the hot corrosion resistance increases, but at the expense of the creep strength. The best combination of the two properties at a given Cr level is achieved in single crystals. The current alloys of choice by the GT manufacturers seem to favor lower Cr content SC alloys to achieve high creep strength. Clearly improved hot corrosion resistance in these alloys would be desirable. Some of the key improvements needed in SC technology are as follows.

- Current SC alloys are too low in Cr and not hot corrosion resistant enough.
- Large thermal gradients (7°C/mm) are needed in the casting process because of the large size. Insufficient gradients lead to (1) casting defects such as equiaxed grains, freckles, low angle boundaries; (2) porosity and reduced fatigue strength; (3) large dendrite spacings which are difficult to solutionize. Lower growth rates can counter these effects, but decrease output and increase cost.
- Need to cast single blades gives low throughput.
- To accommodate large gradients, high superheat temperature (> 1500°C) and metal mold reactions-casting parameters need to be optimized.
- Need defect acceptance criteria.

In summary, cost effective, high quality SC manufacturing technology is needed.

2.5 Coatings

The evolution of coatings for blades and vanes have tracked the need to address different forms of corrosion with increasing temperature, coupled with a need to withstand thermal cycles without spalling or cracking. Three forms of corrosion have been identified as shown in Figure 6(18). A layer type corrosion characterized by an uneven base metal-oxide interface and the absence of subscale sulfides, known as type II hot corrosion occurs below about 700°C (1300°F). A non layer type corrosion (type I) characterized by a smooth base metal-oxide interface and a continuous, uniform precipitate depleted zone containing discrete sulfide particles beneath the oxide scale has been found to occur above 775°C (1425°F). Above approximately 1700°F, oxidation takes over as the primary corrosion mechanism. The earliest coatings were aluminides applied by pack cementation and used primarily to protect against oxidation. With increasing evidence of type I hot corrosion in the late '60s, Pt-aluminide coatings such as RT22 and RT44 were developed to provide greater resistance to type I corrosion and to thermal cycling. In the mid '70s, type II hot corrosion became a common problem and required chromium additions to the coating to combat the problem. Since Cr additions to aluminide coatings could not be implemented by pack cementation, new overlay coatings based on CoCrAlY were applied by electron beam physical vapor deposition (EB-PVD). The CoCrAlY compositions developed initially required further optimization for type II corrosion and thermal cycling resistance. Addition of nickel, and increasing the Cr/Al ratio resulted in a new class of NiCoCrAlY and CoCrAlY compositions, applied by plasma spray techniques. Some current overlay coatings contain a double layer, an outer aluminide layer for oxidation resistance and an inner layer of CoCrAlY or NiCoCrAlY balanced for type I and II hot corrosion. Several of the overlay coatings also contain hafnium, tantalum and rhenium in some combination with yttrium. Several coatings also contain trace amounts of Si to improve coating adhesion.

In addition to the corrosion resistance required of the coating, its mechanical properties are also of importance. Virtually all the coatings are based on an aluminide intermetallic phase, and these are brittle at low temperatures and ductile at high temperatures. The temperature at which the change takes place is called the ductile to brittle transition temperature (DBTT). It is important that the DBTT is as low as possible, so that cracking in the coating does not occur in service since the cracks may then propagate into the substrate. Bernstein at al (19) make a most important point: NiCoCrAlY coatings containing 20 to 26% cobalt are significantly more ductile than either NiCrAlY or CoCrAlY coatings. Figure 7 illustrates these points (20-23).

At this time all of the major industrial gas turbine manufacturers use overlay coatings on at least the first two stages of turbine blades in their 1260°C (2300°F) turbines. This type of coating is also scheduled for use in the 1425°C (2600°F) turbines as well, although in some cases the overlay serves a second important role as a bond coat for a ceramic thermal barrier, as discussed in the following section. Compositions of these overlays are often proprietary. Nevertheless, the goals of oxidation and corrosion protection to at least 900°C (1650°F) combined with metallurgical compatibility with the blade alloy and ductility tend to dictate a balance of Ni and Co and of Cr and Al with around 1% of Y and/or some other minor elements. In most cases, the coatings are applied by either a plasma spray or a vapor deposition process.

2.5.1 Thermal Barrier Coatings.

With turbine firing temperatures exceeding 1120°C (2050°F) and thereby challenging the ability of air cooling to maintain surface temperatures below about 900°C (1650°F), there has been a very considerable growth in the use of thermal barrier coatings. The function of the thermal barrier is to retard the conduction of heat from the combustion gas to the internal cooling air, and thus either reduce the cooling air needed, or allow an increase in the turbine inlet temperature for a given metal temperature.

The history and status of TBCs are covered by Bernstein et al (19). They consist of a porous layer of zirconium oxide (ZrO_2) in the cubic modification, which is stabilized by yttria (Y_2O_3) and MgO. Other stabilizers are being examined. They are applied, usually using plasma spray methods, over a bond coat, which is usually an MCrAlY, where M is Ni and/or Co. The bond layer is typically 75-125 μm thick and the oxide layer is 125-375 μm thick. Their function, as the name implies, is to act as a thermal barrier. Because of their low thermal conductivity, they act to decrease the metal surface temperature. The decrease may be as much as 100°C (180°F). Perhaps even more importantly, TBCs smooth out hot spots thus reducing thermal fatigue stresses.

Thermal barrier coatings are currently used in the combustion section (combustor cans/liners and transitions) of some gas turbines, including the newer models from Siemens-Westinghouse and GE. Westinghouse also began using TBCs on the first stage stationary vane in the late 1980s on the upgraded 1120°C (2050°F) machine and continues this application through the 1425°C (2600°F) model which have TBCs applied to the first and second stage blades and vanes (7). Recent development in electron-beam physical vapor deposition of TBCs has resulted in potential improvements over the air plasma spray process for the ceramic layer. This is expected to result in wider use on the airfoils of vanes and blades.

Thermal barrier coatings are considered to be extremely important to the future of 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 aerodynamic and cooling designs of the advanced land-based gas turbine are 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.

2.6 Turbine Disks

The turbine rotating blades are commonly inset into the disks by means of "fir-tree" roots. Rotational forces are thus transmitted through the interconnected disks that comprise the turbine rotor. The disks are highly loaded by the centrifugal force of the rotating blades and would run at temperatures approaching those for blades, unless cooled. The potential for increased disk temperature of course is directly related to the turbine inlet temperatures. As was discussed in a prior section, this presents a significant design challenge. Through their currently announced turbine models, Alstom, Siemens, Siemens-Westinghouse, and MHI have chosen to address this challenge by increasing the cooling applied to conventional disk materials (primarily alloy steels).

GE has elected to introduce nickel-base alloy disks (GE uses the term "wheels" for the turbine disks), together with a lesser amount of air cooling. Schilke et al (5) briefly discuss disk materials in land-based GE gas turbines. They comment that land based turbine wheels have very large diameters and section thicknesses compared to those for aircraft engines. With the increase in firing temperatures and pressure ratios for the advanced land-based engines, GE has chosen Inconel 706 for the turbine disks, first introduced in the production MS7001F. This alloy is similar to Inconel 718 but is easier to produce in the large ingot sizes required for the land-based turbines. The transfer of nickel-base alloys from aircraft engine applications to land based turbines has not been painless. Difficulties have been cited in producing defect free ingots in the large sizes necessary (over 22,000 lb. forgings) (26). Microcracking occurred in some of the early MS7001F disks after 10,000 to 30,000 hours of service. Shot peening is now relied upon to help prevent the microcracking. Recently, processing advances have permitted IN718 to be used for wheels in the H series engines (5a).

3. ADVANCED MATERIALS

3.1 Oxide Dispersion Strengthened Alloys (ODS)

The International Nickel Company developed the mechanical alloying process to produce ODS alloys (25). In this process, mixtures of elemental metal powders, master alloy powders, and very fine refractory oxide particles are prepared and charged into a high-energy ball mill or attritor, where mechanical alloying occurs. 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 used to optimize the strength properties. The most commonly investigated ODS alloy, MA 6000, contains 15% Cr, 2.5 Vol. % Y_2O_3 with various amounts of W, Al, Ti, Fe, Ta, Mo, C, B, and Zr. The lower density of MA 6000 compared to DS Mar-M 247 provides for lighter weight blades; the higher modulus of MA 6000 may, however, be unfavorable since peak load stresses can be generated by smaller transient strains.

It is reported that the ODS alloy MA 6000 has greater rupture strength than several conventional and single crystal superalloys at high temperatures/high stresses (26). The fact that ODS superalloys are better than SC superalloys at high temperatures/low stresses but somewhat worse at low temperatures/high stresses clearly means that some turbine components will benefit from being fabricated from an ODS alloy while some others will not.

Another important property that must be considered when selecting superalloys for turbine blade applications is fatigue strength. Direct SC/ODS comparisons are not available from the literature. Both SC and ODS alloys seem to offer advantages over conventional alloys because of reduced size of crack nucleation sites. Fatigue cracks have been found to nucleate from micropores (50-80 μm) in industrial blade single crystals (27). In ODS alloys stringers of inclusions play a dominant role in crack nucleation (30-60 μm in diameter and more than 100 μm in length in MA 6000). An immediate advantage of one or the other system is not apparent. This is different, however, in strain controlled fatigue where a low modulus is desired. While the orientation of single crystals is always a favorable <100> some ODS alloys like MA 6000 display a high modulus <110> texture. Texture control in ODS alloys by application of specific thermomechanical processing cycles should be possible but has not yet been employed successfully.

If ODS superalloys are to be adopted for industrial gas turbines, similar problems as with single crystals arise. Alloys with adequate corrosion resistance need to be developed. The corrosion resistance of MA 6000 is similar to IN 738, but it is not sufficient to exploit the strength potential of the alloy. Component manufacturing technologies must be scaled up to the larger sizes. Just as for single crystals, ODS superalloy technology relies on a high thermal gradient that needs to be established during zone annealing of the alloy. The ODS alloys currently have the potential for use in combustion section, but they present problems for fabrication by conventional fusion welding techniques because of the agglomeration of the oxides in the overheated zone and the accompanying strength loss. Diffusion bonding, brazing and mechanical joining are, however, possibilities.

3.2 Ceramic Materials

The monolithic ceramics (mainly SiC, Si_3N_4, Al_2O_3) and various silica glass and glass ceramic systems) have been the subject of numerous government and private industry sponsored development and demonstration programs. The major shortcomings of the older generation of ceramics are low fracture toughness, existence of a wide distribution of flaw sizes, including many undetectable flaws, poor thermal shock resistance, susceptibility to slow crack growth at intermediate temperatures of 700 to 1100°C (1292 to 2012°F) and susceptibility to excessive creep at temperatures approaching 1100°C (2012°F). The new generation of silicon nitrides has shown all around improvements in toughness, thermal shock resistance, creep resistance and resistance to slow crack growth. The major problem still appears to be the prediction of failure, and the inability to detect sub-critical cracks. There have been a few "successful" component demonstration tests, but in general there is insufficient data for realistic statistical component failure analysis.

Composite ceramics with particles, platelets, whiskers, or short fibers, all exhibit brittle behavior with zero overload capacity. Therefore, they are treated in the same general class as monolithic ceramics. In the near term monolithic ceramics will most likely be restricted to lightly loaded structures with strain isolation attachment features. High strengths are not achievable unless the matrix makes a significant contribution. In general, the lack of maturity of these systems and their cost make them inappropriate for land based power generation systems at the current time.

3.3 Intermetallics

Although laboratory results are very promising and development work in titanium aluminides is well advanced, no ordered alloy has reached the where sufficient manufacturing and design data exists 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.

In large land-based gas turbines, sizes will be predominantly larger and the advantage of weight saving should be less than in aeronautical equipment, except for the last row of blades. 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.

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.

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Table 1 Evolution of Large Land Based Gas Turbine Design Features

Approx. 1st Year	1967	1972	1979	1990*	2000**
Approx. RIT °C (°F)	900 (1650)	1010 (1850)	1120 (2050)	1260 (2300)	1425 (2600)
Compr. Ratio	10.5	11	14	14.5	19-23
Exhaust Temp. °C (°F)	427 (800)	482 (900)	530 (986)	582 (1080)	593 (1100)
Cooled Turbine Rows	R1 vane	R1&2 vane, R1 blade	R1&2 vane, R1&2 blade	R1,2,3 vane, R1,2,3 blade	R1,2,3 vane, R1,2,3 blade
SC Power Range (MW)	50-60	60-80	70-105	165-240	165-280
Approx. Heat Rate (SC) Btu/kWh	11,600	11,180	10,250	9,500	8,850
Efficiency, %	29	31	34	36	39
Approx. Heat Rate (CC) Btu/kWh	8,000	7,350	7,000	6,400	5,880
Efficiency, %	43	46	49	53	58

* Corresponds approximately to GE 7F, FA/9F, FA, S-W 501F, and MHI 501F/701F conditions; Siemens V84.3/94.3 is in the same class, in terms of output and heat rate.

** Corresponds approximately to GE 7H/9H, S-W 501G, MHI 701G conditions;
Siemens V84.3A/94.3A and Alstom (ABB) GT24/26 are in the same class in terms of output and heat rate.

SC and CC denote Simple Cycle and Combined Cycle.

Table 2 Combustion Turbine Materials (Current Models)

OEM	Combustor	Transition	Turbine Discs
Alstom (ABB)	Annular-Nickel alloy (e.g. Hastelloy X) with ceramic refractory tile	IN 617 (Duct with silo configuration)	Alloy steel (A469/565) (12CrNiMoV) (welded)
GE	Cans-Hastelloy X and Haynes 188 with TBC	Nimonic 263 with TBC	Nickel alloy (IN 706 & IN 718)
Siemens	Annular-Nickel alloy (e.g. Hastelloy X) with ceramic coating	IN 617 (Duct with silo configuration)	Alloy steel (A565) (22CrMoV & 12CrNiMo)
S-W	Cans-Hastelloy X with TBC	IN 617 with TBC Haynes 230 with TBC	Alloy steel (NiCrMoV)
MHI	Cans-Tomilloy with TBC	Tomilloy (similar to IN 617)	Alloy steel (NiCrMoV)

Table 3 Combustion Turbine Alloys and Coatings for Hot Section Blading

OEM	Model	Vanes	Blades	Coatings
Alstom (ABB)	11N2	IN939	IN738LC	NiCrAlY+Si (some rows)
	GT24/GT26	DS CM247LC(R1)/MarM247LC(R2&3)/ IN738) (R4&5)	DS CM247LC(R1-3)/MarM247LC (R4&5)	TBC (R1V)/NiCrAlY+Si (R2-4 B&V) Uncoated (R5V), Chromized (R5B)
GE	7/9EA	FSX-414 (All Stages)	GTD111(R1)/IN738(R2)/U-500(R3)	RT22 →GT29-In+ (R1B)
	7/9FA	FSX-414 (R1)/GTD222 (R2&3)	DS GTD111 (R1)/GTD111 (R2&3)	GT33-In(R1)/GT29-In+(R2)/ Chromized (R3)
	7H	SC Rene N5 (R1)/DS GTD222 (R2) Rene108 (R3)/GTD222 (R4)	SC Rene N5 (R1)/DS GTD111 (R2) DS GTD444 (R3&4) (DS version of Rene N4)	TBC (R1&2 B&V) All others GT 33
Siemens	V84/94.2	IN939 (All 4 rows)	IN738LC (R 1, 2, &3)/IN792 (R4)	CoNiCrAlY+Si (some rows)
	V84/94.3A	(SC)PWA1483 (R1&2)/IN939 (R3&4)	(SC)PWA1483 (R1&2)	TBC (EB-PVD) (R1B)/MCrAlY+Re
S-W	501D5/D5A	ECY-768 (R1&3)/X-45 (R2&4)	IN738 (R1)/U-520 (R2, 3, &4)	TBC (partial on R1V)/MCrAlY (some blade rows)
	501F	ECY-768 (R1, 2, & 3)/X-45 (R4)	IN738LC (Four rows)	TBC (R1 B&V)/MCrAlY (R2&3B)/ Sermalloy J (R4B)*
	501G	IN939 (Four rows)	DS MM002 (R1&2)/EA CM247 (R3&4) Or DS CM 247 (R1&2)	TBC (EB-PVD) (R1 B&V) TBC (R2 B&V) MCrAlY (R3 B&V)
MHI	501/701F	MGA2400 (CC Ni alloy) (Four rows)	MGA1400 CC (Similar to IN792) (Four rows)	TBC (R1 B&V)/MCrAlY (R2&3B)
	501/701G	MGA2400 (CC Ni alloy) (Four rows)	MGA1400 DS (R1&2)/MGA1400 CC (R3&4)	TBC (R1&2 B&V)/MCrAlY (R3 B&V)

Rene N5: 7.5Co, 7Cr, 1.5Mo, 5W, 3Re, 6.5Ta, 6.2Al, 0.05C, 0.2B, 0.01Y.
SC PWA 1483: 12%Cr Alloy similar to INCO792
GT29: CoCrAlY; GT29+: CoCrAlY + Diffusion Aluminide Top Coat
GT33: MCRALY with Improved Oxidation Resistance
* TBC with MCrAlY bond coat for Row 1 Blade & Vane (Plasma Sprayed)
+ TBC with MCrAlY bond coat for Rows 1 and 2 Blades & Vanes

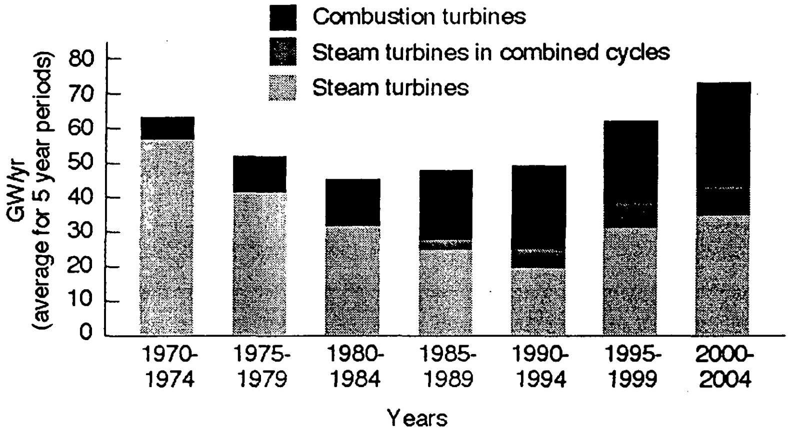
Table 4 Composition (wt%) of Cast DS Superalloys for Turbine Blade Application (Ref 13)

	Composition, wt%												Approx. Yr.	Temp.
Alloy	C	Cr	Co	Al	Ti	Mo	W	Nb	Ta	Zr	B	Other	of Intro-duction	Capability (°C)
Directionally Solidified														
DSGTD 111	0.10	14.0	9.5	3.0	4.9	1.5	3.8		2.8	0.03	0.01		1987	
DSMAR M247		8.25	10	5.5	1.0	0.78	10		2.8		0.015	1.5Hf		
MAR-M200 Hf	0.15	9.0	10	5.0	2.0	–	12.5	1.0	–	0.05	0.02	2.0Hf	1970	1040
MAR-M002 DS	0.15	9.0	10	5.5	1.5	–	10	–	2.5	0.05	0.015	1.5Hf	1975	1045
IN-6203[2]	0.15	22	19	2.3	3.5	–	2.0	0.8	1.1	0.1	0.01	0.75Hf	1981	1020

Table 5 Nominal Compositions of Three Generations of Single-Crystal Superalloys (wt. %) (Based on Ref 16)

Alloy	Cr	Co	Mo	W	Ta	Re	V	Nb	Al	Ti	Hf	Ni	Density (kg/dm^3)	Year	Temp Capa-bility, °C*
First Generation															
PWA 1480	10	5	–	4	12	–	–	–	5.0	1.5	–	Bal.	8.70	1980	1060
PWA 1483	12.8	9	1.9	3.8	4	–	–	–	3.6	4.0	–	Bal.	–	–	–
René N4	9	8	2	6	4	–	–	0.5	3.7	4.2	–	Bal.	8.56	–	–
SRR 99	8	5	–	10	3	–	–	–	5.5	2.2	–	Bal.	8.56	1980	1080
RR 2000	10	15	3	–	–	–		–	5.5	4.0	–	Bal.	7.87	–	–
AM1	8	6	2	6	9	–	–	–	5.2	1.2	–	Bal.	8.59	1985	1090
AM3	8	6	2	5	4	–	–	–	6.0	2.0	–	Bal.	8.25	–	–
CMSX-2®	8	5	0.6	8	6	–	–	–	5.6	1.0	–	Bal.	8.56	1980	1070
CMSX-3®	8	5	0.6	8	6	–	–	–	5.6	1.0	0.1	Bal.	8.56	–	–
CMSX-6®	10	5	3	–	2	–	–	–	4.8	4.7	0.1	Bal.	7.98		
CMSX®-11B	12.5	7	0.5	5	5	–	–	0.1	3.6	4.2	0.04	Bal.	8.44		
CMSX®-11C	14.9	3	0.4	4.5	5	–	–	0.1	3.4	4.2	0.04	Bal.	8.36		
AF 56 (SX 792)	12	8	2	4	5	–	–	–	3.4	4.2	–	Bal.	8.25		
SC 16	16	–	3	–	3.5	–	–	–	3.5	3.5	–	Bal.	8.21	1990	1030
Second Generation															
CMSX-4®	6.5	9	0.6	6	6.5	3	–	–	5.6	1.0	0.1	Bal.	8.70	1986	1110
PWA 1484	5	10	2	6	9	3	–	–	5.6	–	0.1	Bal.	8.95	1986	1100
SC 180	5	10	2	5	8.5	3	–	–	5.2	1.0	0.1	Bal.	8.84		
MC2	8	5	2	8	6	–	–	–	5.0	1.5	–	Bal.	8.63	1990	1125
René N5	7	8	2	5	7	3	–	–	6.2	–	0.2	Bal.	NA		
Third Generation															
CMSX®-10	2	3	.4	5	8	6	–	.1	5.7	.2	.03	Bal.	9.05		
René N6	4.2	12.5	1.4	6	7.2	5.4	–	–	5.75	–	.15	Bal.	8.98		
CM186LC	6	9		8	3	3			6	0.7	1.4	Bal.		1996	

*Based on 100 h to rupture at 140 MPa
NASAIR 100, CMSX-2 and CMSX-3 are derivatives of Mar M247.



Source: SFA Pacific from Siemans Data, cited in Ref. 1

Figure 1
Worldwide Fossil-Fueled Power Plant Orders for Combustion and Steam Turbines

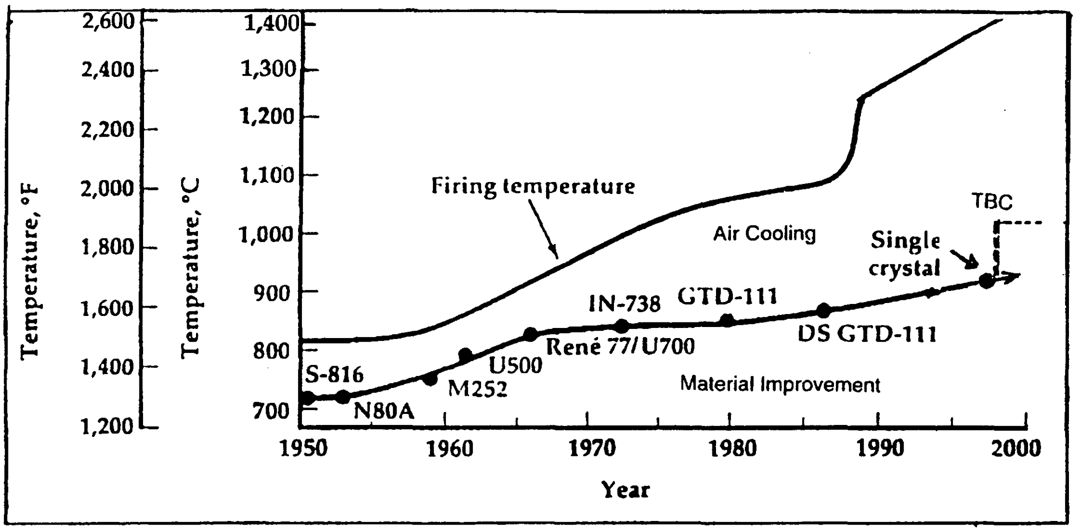


Figure 2
Trends in turbine firing temperature and bucket-alloy capability. The sharp rise in firing temperature reflects the advent of advanced air-cooling technology. Alloy temperature capability is in terms of rupture temperature for 100,000 hr at 140 MPa (20 × 10^3 psi). (Based on adaptation of Ref. 5)

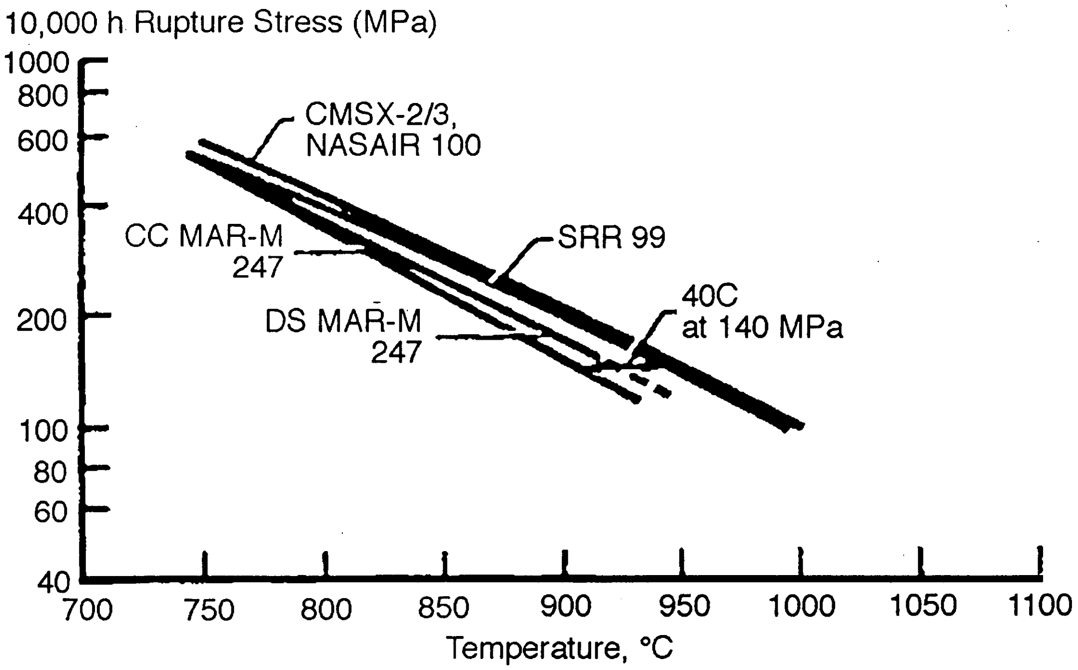


Figure 3
10,000 h Rupture Strength of the SC Superalloys CMSX-2/3, NASAIR 100 and SRR 99 as Compared to Conventional CC and DS Material.

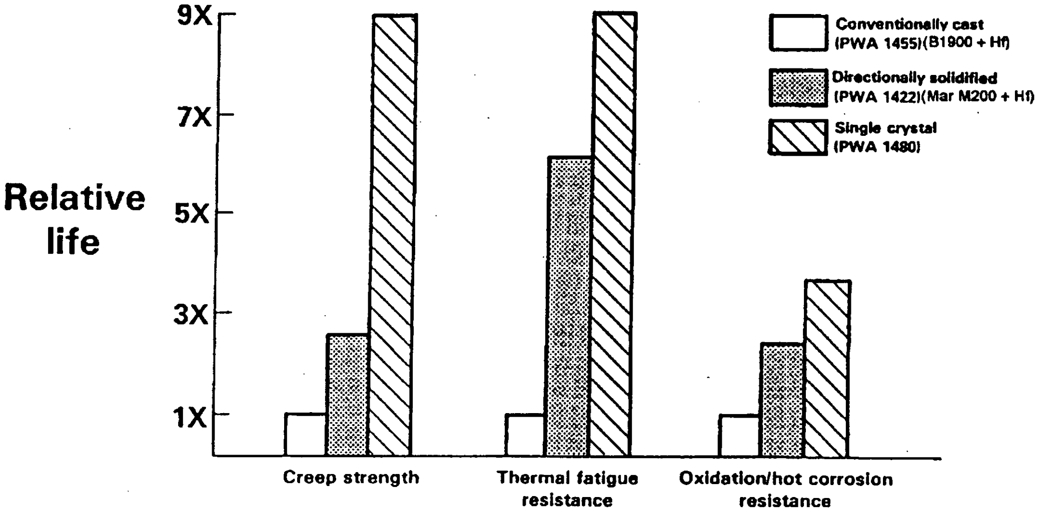


Figure 4
Major Testing Program Demonstrates Single Crystal Benefits (Ref. 18)

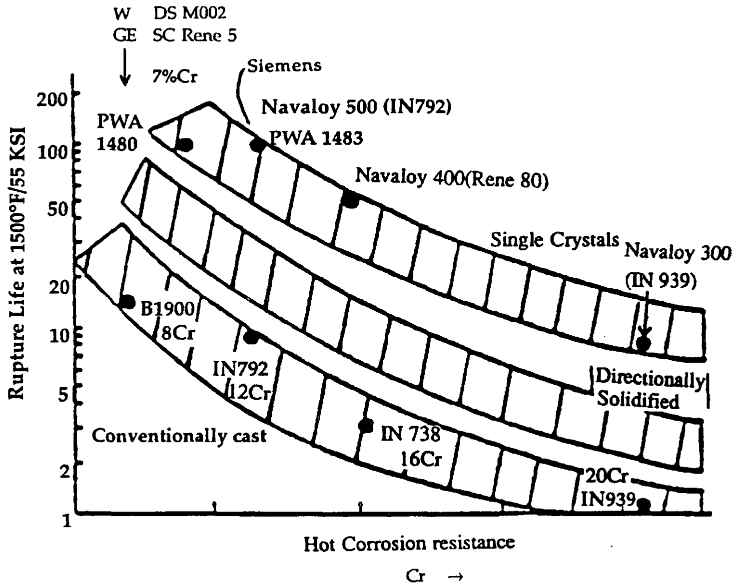


Figure 5
Single Crystals provide significant improvement in terms of strength/hot corrosion combination
- Hot corrosion resistance (i.e. Cr) and creep strength move in opposite directions.
- Best compromise achieved in SC alloys.

(Ref. 18)

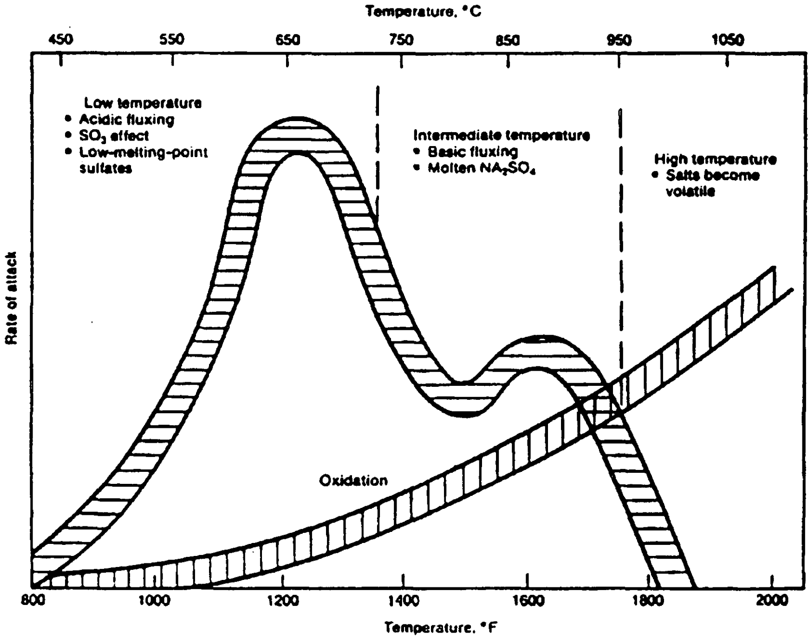


Figure 6
Schematic illustration of the variation in corrosion rate with temperature due to changes in hot-corrosion mechanism (Ref. 18, as cited in Ref. 4).

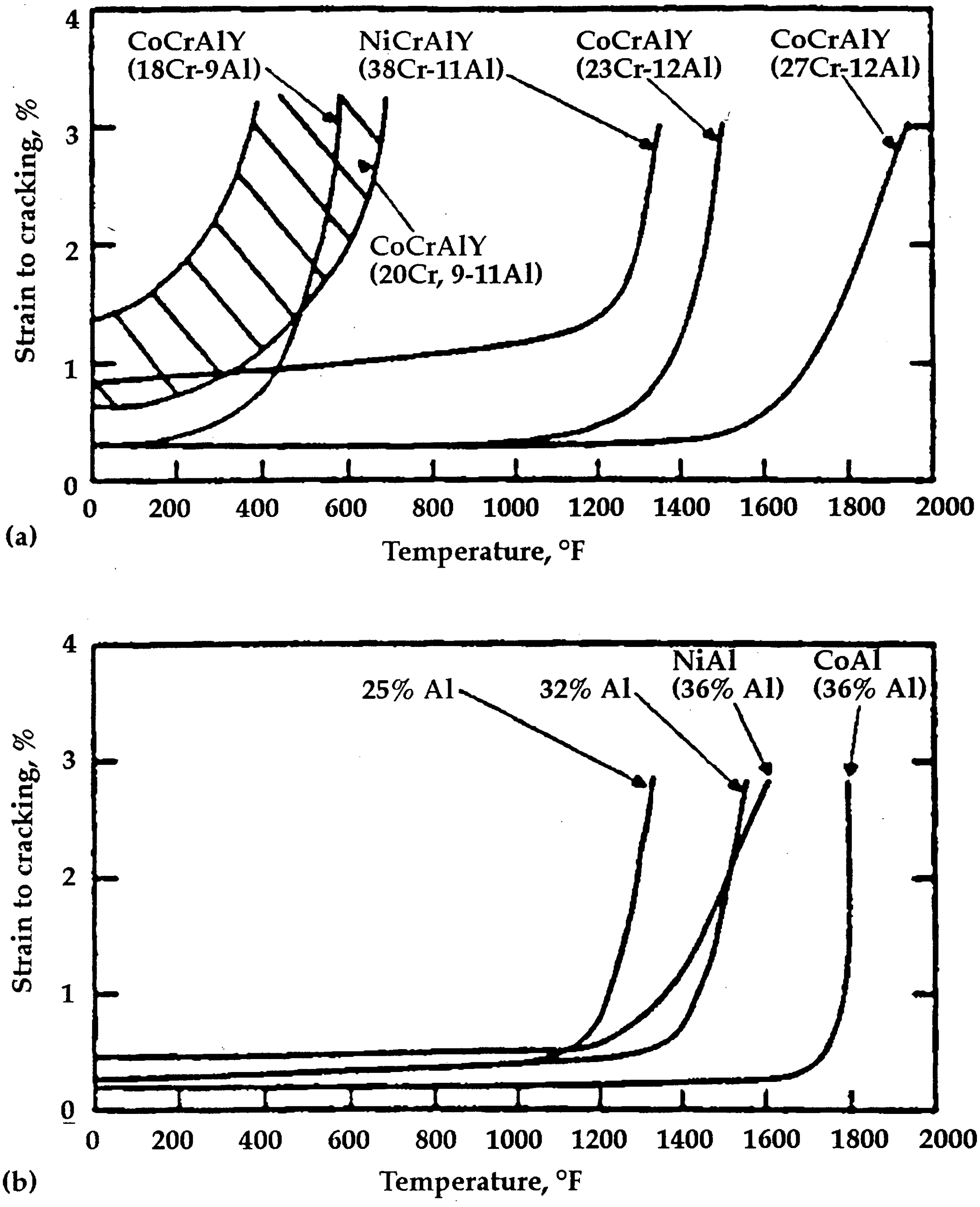


Figure 7
Strain to cracking as a function of temperature. (a) Overlay Coatings, (b) Aluminide Coatings. [21]