This is the third of a four-part series about development of materials for gas turbine aircraft engines. This article covers alloy development, investment casting, and high-temperature coatings.

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Efficient gas turbine engines for modern jet aircraft are possible only because of the materials that have enabled continuous improvement in high-temperature operation, higher power, and reduced weight over the past 50 years. This is the third of a four-part series about development of these materials.

The materials challenges presented by gas turbine engines are particularly daunting given the high stresses, high temperatures, and high reliability demanded of these engines and their components. This article discusses superalloy development, investment casting, melting process improvements, and high-temperature coatings.

Superalloy development
By the late 1950’s, turbine engine designers had become limited by the predicted mechanical and temperature constraints of the stainless steels. The well-known fabrication difficulties of the then-current nickel-base superalloys also prevented advances in engine development. These difficulties included strain-age cracking during post-weld heat treatment, and cracking during forging.

Driven by the U.S. Department of Defense commitment to aerospace, manifold government funding was made available to universities and national and industrial laboratories for alloy and process improvement. Alloy producers forecasted market growth based on expectations of DOD advanced systems, a growing commercial aircraft market, and the NASA commitment to space systems. Consequently, alloys emerged from laboratories at an unprecedented pace.

Significant alloying and process improvements were made on the gamma-prime strengthened nickel-base systems. These new alloys were stronger at all temperatures of interest, and were capable of service at much higher temperatures.

At the same time, progress was made on the iron-nickel base alloys strengthened by gamma-double-prime (γ″, Ni3Nb). Although they do not have quite the temperature capability of their elder gamma-prime cousins, they possess high tensile strength and are more easily processed and welded. Their sensitivity to notch failure in creep rupture tests was readily corrected via improvements in melt processing and sulfur control.

These alloys were exemplified by Inconel alloy IN718. The invention of IN718 was a seminal event in enabling modern turbine engines; it is arguably the most successful superalloy ever.
IN718 quickly found adherents within the propulsion industry because of its excellent balance of properties, reasonable cost, and its castability and forgeability. Even today, nearly 50 years after the introduction of IN718, it is the material of choice for the great majority of engine components in applications below approximately 650°C (1200°F).

The counterpart in titanium alloys is the now-standard titanium alloy, Ti-6Al-4V. This was also developed during early 1950s, and is still the workhorse of the industry. Figure 1 shows the percentage breakout of metals in CF6 engine forgings as of the year 2000: IN718 accounted for one-third of the finished engine weight.

Other alloy advances

Many factors contributed to the evolution of today’s alloys, of which many are tailored for specific-use domains. The chronological description of rapid advancements in superalloys during this period is difficult to concisely capture. The sequence of discoveries of the effects of key alloying elements and the development of useful heat treatments were intertwined, and were often made independently in different locations. Stated succinctly, progress was rapid, but also chaotic and undisciplined, and often done under the cloak of “company proprietary” research and development.

Some of the significant discoveries that have subsequently guided superalloy evolution include the following:

- Precipitate phases: Excessive alloying additions were found to result in the gradual but persistent formation of topological close packed (TCP) intermetallic compounds at service temperatures. The result is that the alloy becomes embrittled during service exposure by the depletion of key elements from the matrix and by the acicular shape of the TCP phases, which effectively act as notches. Similarly, it was discovered that slow precipitation of phases considered non-embrittling, such as carbides, reduced alloy strength during service life.
- Temperature differential: For mechanical property control of most wrought alloys and many cast alloys, a meaningful difference (generally, 30 to 110°C, 55 to 210°F) between the hardening precipitate phase solvus (dissolution) temperature and the alloy melting point is necessary. This permits re-solutioning and re-precipitation of strengthening phases for mechanical property control.

However, this criterion limits the quantity of strengthening elements that can be added to the matrix, because highly alloyed metals have little temperature separation between the liquidus and solidus temperatures. (As will be discussed next month, this limitation was one of the motivations for the development of powder metallurgy super-alloys.)

- Property tradeoffs: Alloys can be tailored for specific environments, such as improved resistance to oxidation or sulfidation. However, this tailoring causes a reduction in other mechanical and/or physical properties.
- Formability limitations: Alloy compositions possessing the “best” properties for each application were not always producible in the design-required configuration because of inherent formability limitations.
- Cobalt alloys: Cobalt alloys may be strengthened only by solid solution alloying and carbide distribution. Therefore, cobalt alloys do not have the breadth of application possible with nickel alloys. However, cobalt alloys do have several other advantages over high-temperature nickel alloys. They have higher melting points, and therefore higher service temperatures. They also have better environmental resistance in certain environments, and better weldability because they are solution strengthened. These factors make cobalt advantageous for static components that do not require high strength.

Weld repair

In parallel with the alloy improvements, developers made significant improvements in the processes by which superalloy wrought and cast shapes could be produced. The importance of these processing improvements to the design of turbine engine components cannot be over-emphasized. For example, weldability improvements in the nickel and titanium alloys allowed significant increases in the size and complexity of cast structures by enabling weld repair of casting defects. This allowed manufacture of complex-shaped components such as frames with only a few parts, resulting in lower component weight and higher reliability.

Investment casting breakthrough

The manufacturing processes for turbine airfoils has changed dramatically. Early turbine airfoils were produced from wrought alloys; a solid (non-cooled) superalloy turbine airfoil was investment cast for the first time in the 1950s. This was made possible by the development of improved processes which reduced casting defects that limited strength.

Two other forces drove the change from wrought to cast airfoils. First, the airfoils became more complex to accommodate internal cooling air passages. Second, engineers discovered that airfoils could be cast as single crystals, a microstructure that greatly enhanced their creep-rupture life.

As design technology advanced, the relentless push for ever-greater turbine inlet temperatures became more important. A key limitation was the creep-strength capability of turbine blades and
vanes at the higher operating temperatures.

- **Internal convection:** Engineering analysis indicated that engine performance would be significantly improved if the turbine blades were cooled by internal convection and by a thin film of cooler air over the hottest regions of the blade. Because the air would have to be pressurized to pass through the channels and holes, it was necessary to re-direct, or “bleed,” a portion of the pressurized air in the compressor to the turbine blades. Initially, this technology was implemented by drilling internal cooling air passages in turbine blades. However, drilling was very expensive and time consuming.

- **Ceramic cores:** Fortunately, continued advances in investment casting technology (based on the ancient “lost wax” process) soon made wrought turbine blades obsolete. Key to this technology was the development of ceramic core materials that could be placed inside the turbine blade casting mold to produce intricate internal passageways for cooling air. These ceramics had to be strong enough at molten superalloy temperatures to retain their shape, yet they also had to be completely removed afterward without damaging the casting itself.

The breakthrough came with the discovery that ceramic cores could be removed by caustic solutions contained in heated and pressurized autoclaves. This technology became the basis for the evolution of “cast-in” cooling passages.

- **Laser drilling:** In addition, high-energy laser beams were applied to drill the small-diameter holes connecting airfoil surfaces to the cast-in internal passages. This allowed cooling air to escape to form a film of cooler air over the blade surface, as shown in Fig. 2. The film of cooler air contributed to further increasing the durability of turbine blades, particularly at airfoil leading and trailing edges, which typically are subjected to higher temperatures.

Strong turbine airfoil castings that required little or no subsequent machining, even with elaborate internal cooling passages, were finally achieved after overcoming challenges such as alloy stability, ductility compromise, and property balance.

Large structural castings such as engine casings were also improved. The driving forces were reduced cost and improved reliability through production of a monolithic structure. Although large structural castings are not subject to an environment as demanding as that for airfoils, their large size and complex tooling add unique challenges.

**Coatings versus ductility**

The progress made in alloy development and investment casting technology provided significant improvement in turbine blade performance. However, coating technology also had to be improved for these benefits to be fully exploited.

As alloy compositions continued to achieve ever greater high-temperature strength, environmental degradation (primarily oxidation and sulfidation) of uncoated airfoil surfaces, including internal cooling passages, became a leading systems concern regarding deterioration and cost.

In response to this need, oxidation-resistant coatings were developed and applied to airfoils to improve environmental resistance. Initial coatings were “pack” coatings, which diffuse aluminum into the airfoil surface, forming an adherent and oxidation-resistant layer of NiAl, typically 2 to 3 mils in thickness. However, this improvement in oxidation resistance caused a loss of surface ductility due to the inherent coating brittleness. Design engineers learned to

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**Thermodynamic stability**

It is a challenge to maintain thermodynamic stability of complex, multi-component nickel-base superalloys during extended high-temperature service. In the case of superalloys, super-saturation promotes the formation of brittle intermetallic phases, such as Laves or sigma phases. These severely degrade long-term mechanical properties and cause loss of ductility. In the extreme, they lead to brittle fracture.

Dr. Chester Sims of General Electric and other alloy developers recognized this limitation and introduced models that predicted intermetallic phase precipitation. These rules became part of the alloy developers’ notebooks and enabled them to minimize the problem. However, the breadth of alloy development was significantly constrained by this phenomenon.

- **Ductility:** As more strengthening elements were added, the ductility of the resultant alloys was concomitantly reduced. A major barrier was found at the minimum ductility point for turbine airfoil alloys, usually between 650 and 760°C (1200 and 1400°F). This problem was manifested by the creep-rupture failure of turbine blade attachments, followed by secondary debris that caused downstream damage to the aft turbine stages.

This problem was also manifested by strain-induced severe cracking (even fracture) of turbine airfoils during solidification and/or heat treatment. Engineers adopted a design rule that required airfoil alloys to have at least 2% ductility at 760°C (1400°F). This limited the use of certain high-creep-strength alloys, but was necessary to protect against structural failure. Eventually, the grain boundary ductility shortfall in high-strength blade alloys was solved by an unexpected casting breakthrough: directional solidification virtually eliminated grain boundaries in the principal stress direction.

- **Hafnium oxide:** Fortunately, it was discovered that the addition of hafnium allowed the 2% design-acceptable ductility minimum to be achieved for most alloys. Hafnium readily forms oxides in the melt. In addition, the density of hafnium oxide (hafnia) closely approximates that of the liquid metal, and thus it does not float to the surface. Consequently, hafnia is readily entrapped in the casting, particularly in areas of section size change. These transition zones are often locations of high stress, such as the blade airfoil-to-root transition zone.

- **Property imbalance:** Continued increases in turbine temperature resulted in mechanical and physical property imbalances when it became apparent that the environmental resistance of highly capable turbine airfoil alloys did not match the design strength capability. This was attributable to several factors. First, the intrinsic environmental resistance of the alloys decays with escalation in operating temperature; second, the alloying elements that impart creep resistance often compromise environmental properties; and third, reactive elements that provide environmental resistance perversely encourage the formation of brittle intermetallic phases. Hence, successful alloy developers had to balance a number of complex factors, and the ultimate outcome was (and is) difficult to predict a priori.
adjust airfoil designs to accommodate this fatigue penalty.

Practices were developed to apply an aluminide coating by gaseous phase transport to coat internal surfaces. In one important development, process engineers devised methods to mask the highly loaded areas that should not be coated, such as dovetail attachments, thereby mitigating against premature fatigue failure.

Coating technology developed somewhat autonomously from high-temperature alloy design. What began serendipitously soon became critical, as coating technology became essential to the maturation of high-temperature airfoil designs. Coated alloys (components) with film cooling, were now capable of operating at local temperatures close to the incipient melting point, and at bulk temperatures nearing the dissolution temperature of the hardening-phase (gamma prime). Further hot-section airfoil temperature advances demanded a new concept to gain further capability without the need for even more cooling air.

**High-temperature coatings**

The maximum gas temperature within a jet engine generally is reached during a transient condition. For example, commercial aircraft attain maximum temperatures only at take-off. Consequently, a means for protecting the metal during these transients would allow reduction in cooling air requirements.

Fortuitously, certain ceramics have thermal insulating capability of interest for this transitory condition. The concept of applying a ceramic thermal barrier coating (TBC) might appear absurd initially. One concern was that a ceramic material could not be expected to adhere to the metallic substrate, particularly after repeated thermal cycling, since the coefficients of thermal expansion of the coating and the substrate are different. The variable strain at differing airfoil elements at cooling hole locations further confounds this issue.

Another concern was that the brittle ceramic layer would reduce fatigue life of the airfoil. However, the prior experience of TBCs on static components, notably combustion (burner) liners, and the knowledge that analytical methods could be developed to deal with nil ductility coatings, suggested that success was possible.

These obstacles were overcome, and TBCs based on yttria-stabilized zirconia were successfully developed and applied to airfoils in thin layers, typically 5 to 10 mils thick. Conventional coating processes, such as plasma spray and physical vapor deposition (PVD), were suitable for depositing high quality, thin ceramic layers.

The deposition process was further refined to avoid depositing TBCs in local areas (through direct masking or virtual masking — manipulation of the component-holding device during the coating sequence). The technology also allows depositing variable thicknesses at selected locations. In addition to the thermal management benefit, airfoil design engineers learned to account for the weight of the coating in determining the stress in the blade, and to account for the coating thickness and surface finish in their aerodynamic analysis. Fortunately, the metallic oxidation-resistant coatings, which were applied directly to the surface of the superalloy, also served as an intermediate layer for bonding the

**Overlay coatings**

Spurred by demand for higher engine performance, metal surface temperatures continued to unrelentingly increase. Newer high strength alloys had poorer environmental resistance. Mitigation strategies included further alloy advances and aggressive airfoil cooling configurations.

In response, a family of superior coatings was developed and implemented. These “overlay” coatings were highly oxidation- and corrosion-resistant alloys containing chromium, aluminum, and yttrium, plus solid-solution strengtheners such as cobalt and/or nickel. In line with their elemental composition, they were designated MCrAlY (cobalt and nickel comprise the “M”). They were applied by several different methods, such as plasma spray, electron beam vapor deposition, and sputtering. Often, these coatings were used in conjunction with diffusion coatings to achieve enhanced resistance to deterioration by spalling.

These coatings now serve in a multitude of applications. Initial applications were to turbine blades and vanes in ship propulsion and other aggressive terrestrial environments, such as power plants in coastal areas where sulfidation, as opposed to oxidation, was a leading cause of surface degradation. The technology required to strip and re-coat diffusion and overlay coatings was developed and widely practiced, enabling re-use of what had become very costly superalloy airfoils.

![Fig. 3— Typical Airfoil metallographic cross-section with a TBC coating system. The cross-section of the coated turbine airfoil was taken normal to its axial length. Note distinctive crystallites that comprise the thin ceramic layer; these provide a degree of strain tolerance.](image-url)
ceramic to superalloys. Figure 3 shows a cross-section of a turbine blade with a bond coat and TBC top coat.

The concept of “prime reliability” was soon born. This requires assurance that the thermal barrier coating will not fail and lead to rapid blade loss. However, prime-reliable TBCs have not yet been achieved. Design engineers compensate for the probability of TBC failure by assuming defects will develop. Even with this design margin, application of TBCs has produced more performance gain, an approximate 50°C (100°F) increase in high-pressure turbine gas temperature. This is more than could have been expected from what was initially considered to be technically improbable. Further gains are likely in the drive toward prime reliability.

Eliminating grain boundaries

For most turbine airfoils, the principal stress direction is parallel to the airfoil length. Therefore, it was postulated that if the grain boundaries that serve as crack initiation sites could be aligned with this direction, airfoil life would be greatly enhanced. Spurred by this concept, Dr. Frank VerSnyder of Pratt & Whitney and others began developing the directional solidification (DS) casting process in which the resulting grain boundaries were parallel to the airfoil length. The DS process was implemented as a production process after substantial development effort.

The driving force for DS airfoils was an expected 25°C (50°F) rise in alloy capability. Furthermore, the DS microstructure provided additional benefits, such as much-improved intermediate temperature ductility and greatly enhanced thermal fatigue life. DS also resulted in greater tolerance to local strains and rupture ductility.

This, in turn, made it practical to expect, and therefore to inspect for, local evidence of damage accumulation such as deformation and/or incipient cracking prior to airfoil failure.

The DS process requires longer processing time and greater process control than does conventional (i.e., “equiaxed grain”) casting, and is therefore more costly. The irresistible benefits of the technology culminated in widespread acceptance of DS despite its higher cost.

DS technology was soon followed by the first generation of single crystal (SX) airfoil castings, which eliminated high-angle grain boundaries altogether, with further improvement in stress rupture life and another 25°C increase in allowable blade operating temperature. Figure 4 graphically shows the grain structure for an equiaxed, directionally solidified, and single crystal airfoil.

The oft-used gas turbine milieu term “forgiveness” implies a characteristic that allows damage to be detected significantly prior to component fracture. DS and SX added a degree of forgiveness to high-temperature airfoils, and continue to be a prevention mechanism for catastrophic airfoil failure.

In today’s modern turbine engine, nearly all the high-pressure turbine airfoils are either directionally solidified or single crystals (DS or SX). So compelling is the cost/benefit ratio that the technology is now applied in the even more cost-sensitive stationary gas power generators. Single-crystal castings weighing up to 20 kg are commonly produced for these applications.

For more information:

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