Directionally Solidified and Single-Crystal Superalloys

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THE PRIMARY GOALS in the continuing development of the aircraft gas turbine are increasing operating temperatures and improved efficiencies. A more efficient turbine is required to achieve lower fuel consumption. Higher turbine inlet temperature and increased stage loading result in fewer parts, shorter engine lengths, and reduced weight. Engine operating costs can be reduced if higher temperatures are possible without increasing part life-cycle costs.

Critical turbine components include high-pressure turbine blades, vanes, and disks. During the last 15 years, turbine inlet temperatures have increased by 278 °C (500 °F). About half of this increase is due to a more efficient design for the air cooling of turbine blades and vanes, while the other half is due to improved superalloys and casting processes (Ref 1). The cooling that is now possible with serpentine cores and multiple shaped-hole film cooling (Fig. 1) enables high-pressure turbine blades and vanes to operate with turbine inlet temperatures of typically 1343 °C (2450 °F), which is above the melting point of the superalloy materials. Turbine inlet temperatures as high as 1571 °C (2860 °F) are current parameters for several advanced fighter engines (Ref 2). It is forecast that by the mid to late 1990s, fabricated single-crystal airfoils with ultra-efficient transpiration cooling schemes will be capable of operating in gas temperatures greater than 1649 °C (3000 °F) with sufficient durability and reliability for man-rated flight turbine engines. These advanced superalloy single-crystal diffusion-bonded airfoils will present a major challenge to the emerging ceramic composite component technology.

For the past 28 years, high-pressure turbine blades and vanes have been made from cast nickel-base superalloys. The higher-strength alloys are hardened by a combination of approximately 60 vol% γ′-Ni(Al,Ti) precipitated in a γ matrix, with solid-solution strengthening provided by the powerful strengtheners tantalum, tungsten, and molybdenum. The γ phase, which has an ordered face-centered cubic structure, is coherent with the γ matrix, their lattice parameters being almost identical (<1% mismatch). This allows homogeneous nucleation of the precipitate with low surface energy and long-time stability at temperature, ensuring the potential usefulness of the alloys at elevated temperatures up to 0.85 Tm (melting point) for extended periods of time. Tantalum, tungsten, and hafnium substitute for some of the aluminum and titanium in the γ′, thus stiffening this phase because of their relatively large atomic size. Initially, the blades were made as isotropic polycrystal or equiaxed castings. Under aerospace turbine engine operating conditions, failure of these equiaxed-grain components usually occurred at the grain boundaries from a combination of creep, thermal fatigue, and oxidation.

Development of the directional solidification (DS) casting process to produce blades and vanes with low-modulus (100)-oriented columnar grains aligned parallel to the longitudinal, or principal-stress, axis (Fig. 2) resulted in significant improvements in creep strength and ductility as well as in thermal fatigue resistance (5× improvement). Pratt and Whitney Aircraft (PWA) pioneered this process (Ref 3, 4), as well as its turbine engine application, and has accumulated 18 years of production experience with over 25 million flight hours with DS blades and vanes (Ref 5).

There has been recent interest in DS blades not only for small- to medium-size airfoils for industrial turbines that burn natural gas but also for large base-load electricity-generating machines. Improved fuel efficiency requirements, along with the desire for high-temperature exhaust gases from the gas turbine (to produce steam suitable for co-generation electricity production), have resulted in the development and application engineering of DS blades with component lengths in the range of 305 to 635 mm (12 to 25 in.).
Single-crystal alloy PWA 1480 (Table 1) offered a 25 to 50 °C (45 to 90 °F) temperature capability improvement in terms of time-to-1% creep, compared to the extensively used DS MAR-M 200 Hf alloy (Ref 5). The creep property improvement, which increases with temperature, depended on optimized single-crystal microstructures with full solutioning of the α+β′ phase. The PWA 1480 alloy was developed to utilize the relatively low thermal gradient, single-crystal casting facilities already available as DS production units, without the freckling problems of alloy 444 (single-crystal MAR-M 200 with no carbon, boron, hafnium, zirconium, or cobalt) (Ref 5). Alloy PWA 1480, with its high tantalum (12%) and low tungsten (4%) contents, proved to be unique with this castability feature. Multiple homogenization/solutioning treatments with tight temperature control were developed to completely solution the γ′ in PWA 1480 without inducing incipient melting. Since 1982, PWA has had more than 5 million flight hours of successful experience using turbine blade and vane parts of single-crystal alloy PWA 1480 in commercial and military engines (Ref 8).

Directionally solidified and single-crystal superalloys and process technology are contributing to significant advances in turbine engine efficiency and durability. Furthermore, appreciable gains are forecast, particularly from single-crystal technology over the next 10 years. These gains are expected to arise from the development of higher creep strength and improved oxidation-resistant SX alloy compositions as well as from the development of SX casting and fabrication technology to utilize advanced transpiration-cooling schemes.

**Directionally Solidified Superalloys**

**Chemistry and DS Castability.** Early work with directionally solidified columnar-grain turbine blades in the 1960s involved the superalloys used for conventionally cast, equiaxed blades containing approximately 60 vol% γ′, such as IN 100 and MAR-M 200. The problems encountered ranged from little longitudinal stress-rupture improvement with IN 100 to the lack of transverse ductility and DS grain-boundary cracking with MAR-M 200.

Pioneering work by Martin Metals resulted in the addition of hafnium to conventionally cast equiaxed superalloys to improve 760 °C (1400 °F) stress-rupture ductility and castability. For directional solidification, PWA added hafnium to MAR-M 200, which reduced DS grain-boundary cracking and increased transverse ductility. Although hafnium levels of up to 2% and greater in MAR-M 200 Hf combat DS grain-boundary cracking, increasing levels of hafnium also increased the DS airfoil component rejection rate and the number of quality assurance problems. This was due to the occurrence of HFO inclusions that usually resulted from hafnium-ceramic reactions (core, shell-mold). Other first-generation DS alloys that were successfully and extensively adopted by turbine engine companies included René 80H (René 80 + Hf) by GE, MAR-M 002 by Rolls-Royce, and MAR-M 247 by Garrett. Both MAR-M 002 and MAR-M 247 were originally developed by Martin Metals to contain hafnium for optimized equiaxed turbine blade mechanical properties and castability. The nominal compositions of these first-generation DS superalloys are listed in Table 2. Directionally solidified superalloy turbine blades employed in large commercial turbofan engines for long-distance flights have been used for up to 15,000 h with high reliability.

Continuous improvements in airfoil cooling techniques have usually led to significant gains in gas turbine operating efficiencies. However, these cooling techniques often result in very complex cored, thin-wall (0.5 to 1 mm, or 0.02 to 0.04 in.) airfoil...
Table 3 Second-generation DS and SX superalloys

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Cr</th>
<th>Co</th>
<th>Mo</th>
<th>W</th>
<th>Ta</th>
<th>Re</th>
<th>Al</th>
<th>Ti</th>
<th>B</th>
<th>Zr</th>
<th>Hf</th>
<th>Ni</th>
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<tr>
<td>DS alloy</td>
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<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
<td>0.015</td>
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<td></td>
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<td></td>
</tr>
<tr>
<td>CM 247 LC</td>
<td>0.07</td>
<td>8</td>
<td>9</td>
<td>0.5</td>
<td>10</td>
<td>3.2</td>
<td></td>
<td></td>
<td>5.6</td>
<td>0.7</td>
<td></td>
<td></td>
<td></td>
<td>8.54</td>
</tr>
<tr>
<td>SX alloys</td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
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<td></td>
<td></td>
</tr>
<tr>
<td>PWA 1484 (Ref 8)</td>
<td>5</td>
<td>10</td>
<td>2</td>
<td>6</td>
<td>9</td>
<td>3</td>
<td>5.6</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>8.95</td>
</tr>
<tr>
<td>CMSX-4 (Ref 10)</td>
<td>6</td>
<td>9</td>
<td>0.6</td>
<td>6</td>
<td>7</td>
<td>3</td>
<td>5.6</td>
<td>1.0</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>8.70</td>
</tr>
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</table>

Designs, which can be susceptible to grain-boundary cracking during the DS casting of high-creep-strength alloys, particularly with modern high-thermal-gradient casting processes. Thus, the need for improved DS castability resulted in the development by the Cannon-Muskegon Corporation of CM 247 LC, a second-generation alloy from the MAR-M 247 composition (Ref 9). The nominal composition of this superalloy, which is also known as Rene 108, is given in Table 3. The CM 247 LC alloy has particularly excellent resistance to DS grain-boundary cracking and is capable of essentially 100% γ' solutioning to maximize creep strength without incipient melting or deleterious M₃C platelet formation upon subsequent high-temperature stress exposure, but with adequate transverse ductility retention.

With respect to DS grain-boundary cracking, zirconium and silicon are generally known to be bad actors. Small amounts of a brittle, hafnium-rich eutectic phase containing high concentrations of zirconium and silicon have been found in DS crack-prone tests (Ref 9). It was observed that very small reductions in zirconium and titanium contents, combined with a very tight control of silicon and sulfur, dramatically reduced the DS grain-boundary cracking tendency of a high-creep-strength superalloy such as MAR-M 247 (Ref 11). The major microstructural effect of the lower titanium content in CM 247 LC, compared to MAR-M 247, is to significantly reduce the size of the γ'/γ' eutectic nodules as well as to lower the volume fraction of the eutectic from approximately 4 vol% in MAR-M 247 to 3 vol% in CM 247 LC DS components. This factor is also believed to be significant in reducing the DS grain-boundary cracking tendency of CM 247 LC.

Heat Treatment and Mechanical Properties. Multistep solutioning techniques based on a slow temperature increase between steps and temperatures up to 1254 °C (2290 °F) are used to supersolution heat treat CM 247 LC DS airfoil components to attain microstructures such as those shown in Fig. 4. Resultant stress-rupture property improvements are illustrated in Fig. 5.

The advent of single-crystal technology is not likely to preempt the need for DS airfoil components in the intermediate term. Directionally solidified airfoils will continue to be used for vane segments and low-pressure
blades in advanced turbine engines because of the producibility of the components, which makes them cost effective.

Several third-generation DS superalloys containing rhenium have been developed that have stress-rupture strength values close to those of the first-generation single-crystal alloys (Ref 12). These new alloys are particularly useful for DS vanes where load-bearing capability, such as to support a bearing, is an important design consideration.

Single-Crystal Superalloys

The greatest advance in the metal temperature capability of turbine blades in the last 25 years has been the single-crystal superalloy and process technology pioneered by PWA (Fig. 6 and 7). The dramatic improvement in the durability of the F-100 fighter engine turbine, as evidenced by the service performance of the F-220 version, is largely due to the PWA 1480 superalloy single-crystal first- and second-stage blades and vanes (Fig. 8).

Other pioneering single-crystal alloy development work resulted in the derivation of several single-crystal compositions from MAR-M 247 during the Garrett/NASA Materials for Advanced Technology Engines (MATE) program, which began in 1977 (Ref 14, 15). The two alloys studied extensively were NASAIR 100 and NASAIR Alloy 3; the latter contained a minor hafnium addition.

The compositions of the first-generation single-crystal superalloys, many of which are being used in turbine engine applications, are shown in Table 1: René N-4 developed by General Electric (Ref 16); SRR 99 and RR 2000 by Rolls-Royce plc (Ref 17); AM1 by the Office National d’Etudes et de Recherches Aerospatiales (ONERA) and Snecma (Ref 18); and CMSX-2, CMSX-3 (Ref 19), and CMSX-6 (Ref 20) by Cannon-Muskegon Corporation. These alloys are characterized by approximately the same creep-rupture strength (density corrected) but have differing SX castabilities, grain qualities, solution heat treatment windows, propensities for recrystallization upon solution treatment, environmental oxidation and hot corrosion properties, and densities. Typical stress-rupture properties are shown in Fig. 9 and 10.

Chemistry and SX Castability

Alloy CMSX-2 was developed in 1979 from the MAR-M 247 composition using some of the experience of the Garrett/NASA MATE program (Ref 14). A multidimensional development approach was used to achieve a high level of balanced properties (Fig. 11). The chemistry modifications to MAR-M 247 to develop CMSX-2 (Table 1) are summarized below with respect to function and objectives:

- Grain-boundary strengthening elements (boron, hafnium, zirconium, and carbon) were removed to achieve a very high incipient melting temperature (1335 °C, or 2435 °F)
- Partial substitution of tantalum for tungsten (CMSX-2 has 6% Ta) for good single-crystal castability, high γ' volume fraction (68%), improved γ' precipitate strength, microstructural stability (freedom from α-tungsten and tungsten, molybdenum-rich μ phases), good oxidation resistance, and coating stability
- Cobalt maintained to increase solid solubility and microstructural stability
- Chemistry balance designed to ensure a wide and practical solution heat treatment temperature range, or window (difference between the γ' solvus and the incipient melting temperature), of at least 22 °C (40 °F)
- Phacomp control of the chemistry of the alloy to avoid the occurrence of deleterious topologically close-packed phases
Figure 12 shows the relative potency of tantalum, tungsten, and molybdenum as solid-solution strengtheners in binary nickel alloys, where tantalum is the most powerful strengthener on an atomic percent basis. An increase in the lattice parameter of the $\gamma$ phase due to alloy additions increases the solid-solution strengthening. Tantalum also partitions strongly to the $\gamma'$ phase, increasing the volume fraction and stiffening the $\gamma'$ due to its relatively large atomic size. The strength of the $\gamma'$ phase is important in superalloys with a high volume fraction of $\gamma'$ (>50%) because $\gamma'$ shearing is the primary strengthening mechanism. With the mean free edge-to-edge distance in the $\gamma$ matrix between the precipitates being smaller than the average precipitate size itself, dislocation shearing of the $\gamma'$ particle is favored over Orowan dislocation looping around the $\gamma'$ particles.

Detailed transmission electron microscopy studies of dislocation movement in cast high-strength superalloys, such as MAR-M 002 (Table 2) and its single-crystal derivative SRR 99 (Table 1), have shown the importance of ensuring that the antiphase boundary (APB) energy is high, so that the stacking fault mode of creep deformation occurs at temperatures up to 850 °C (1562 °F), thus ensuring high creep strength (Ref 17). Tantalum additions raise the APB energy relative to the stacking fault energy (Ref 17), leading to the increased tendency for stacking faults to be formed at lower temperatures.

The CMSX-2 alloy is designed to provide good SX foundry performance because castability is a crucial alloy performance criterion for any complex, thin-wall turbine blade or vane component, a characteristic sometimes given limited attention in alloy design. It affects not only the yield and cost of components but also the defect level and therefore component performance. Single-crystal casting defects of concern are:

- **Freckling**: A spiral of equiaxed grains caused by elemental segregation in the liquid state
- **Slivers**: Moderate-angle grain defects
- **Microporosity**: A uniform distribution of interdendritic micropores
- **Spurious grains**: High-angle grain boundaries
- **Stable oxide inclusions**: $\text{Al}_2\text{O}_3$
- **Carbides**: TiC

The partial substitution of tantalum for tungsten in the CMSX-2 alloy, compared to the MAR-M 247 chemistry, helps overcome the freckling problems inherent in the low-tantalum, high-tungsten single-crystal alloys. The strong $\gamma'$-forming elements, aluminum and titanium, which are also low density, tend to segregate to the last liquid to solidify in the interdendritic spaces during the SX solidification process. This can create density changes and consequential...
flow in the liquid metal close to the solidification front, which can nucleate freckle trails of equiaxed grains. This can occur particularly under conditions of low or changing thermal gradients. Tantalum, which is a strong γ-forming element of high density, also tends to segregate to the last liquid to solidify in the interdendritic spaces and thus evens out these density changes in the liquid, or mushy, zone and reduces freckling tendencies.

Several studies undertaken in the United States, Europe, and Japan confirm that high [N] and [O] levels in single-crystal superalloy ingot adversely affect SX casting grain yield, supporting the importance for low [N] and [O] levels in the master alloy. Carbon, sulfur, and [O] master alloy impurities are shown to transfer nonmetallic inclusions, such as Al₂O₃, (Ti,Ta) C/N, and (Ti,Ta)₂ S, to SX parts (Ref 22). Grain defects can nucleate on these inclusions.

Several second-generation, rhenium-containing, single-crystal superalloys have been developed for turbine engine applications. Two typical compositions are given in Table 3. Rhenium partitions mainly to the γ matrix; this retards coarsening of the γ'-strengthening phase and increases γ/γ' misfit (Ref 23). Atom-probe microanalysis of rhenium-containing modifications of the PWA 1480 and CMSX-2 alloys reveals the occurrence of short-range order in the matrix with small rhenium clusters (~1.0 nm, or 10 Å, in size) detected in the γ in the alloys (Ref 24). The rhenium clusters can act as more efficient obstacles against dislocation movement compared to isolated solute atoms in the γ solid solution; therefore, they play a significant role in improving the creep strength.

The Larson-Miller stress-rupture comparison of CMSX-4 and CMSX-2/3 is shown in Fig. 13. The stress-rupture temperature capability advantage of CMSX-4 over CMSX-2/3 is 27 °C (48 °F) (density corrected) in the 248 MPa/982 °C (36 ksi/1800 °F) region. In the 103 MPa/1121 °C (15 ksi/2050 °F) region, the stress-rupture temperature capability advantage is 30 °C (54 °F) (density corrected). The data also indicate that CMSX-4 has a potential peak-use temperature under stress of at least 1149 °C (2100 °F).

**Single-Crystal Casting Techniques.** A variety of single-crystal airfoil component-casting techniques have been developed to production status around the world in the last 10 years. Most involve a withdrawal-type vacuum induction casting furnace with mold susceptor heating. Cooling plate sizes range in diameter from 140 to 610 mm (5½ to 24 in.). Some of the developed SX casting techniques are presented in Ref 12, 13, 25, and 26.

The modern helicopter engine turbine vane shown in Fig. 14 represents a difficult cored configuration. The large shrouds and core make this vane susceptible to shrinkage, grain nucleation, and recrystallization during solution heat treatment. Single-crystal casting processes developed by the Allison Gas Turbine Division of General Motors Corporation result in high yields for this vane in CMSX-3. Similar yields have been demonstrated with CMSX-4 using the same Allison casting process.

**Single-Crystal Heat Treatment and Microstructures.** With regard to solutioning, the latest multistep ramped cycles developed for single-crystal components are designed to completely solution the γ' and most of the γ'/γ eutectic without incipient melting.

An additional benefit of the high-temperature cycles is the element homogenization effect, as shown in Fig. 15. Alloy CMSX-4, which is solutioned at a maximum temperature of 1321 °C (2410 °F) in commercial vacuum heat treatment furnaces, readily attains the 99%+ (<1% remnant γ'/γ eutectic) solutioned microstructure, as illustrated in Fig. 16.

With regard to aging, the weight fraction of γ' in CMSX-2 is approximately 68% with chemistry, as shown in Table 4, both being
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Fig. 11 CMSX-2 alloy development goal

Fig. 12 Influence of alloying elements on the lattice parameter of binary nickel alloys. Source: Ref 21

Table 4 Chemical composition of the γ' phase in CMSX-2

<table>
<thead>
<tr>
<th>Element</th>
<th>Composition, wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nickel</td>
<td>69.25</td>
</tr>
<tr>
<td>Cobalt</td>
<td>3.15</td>
</tr>
<tr>
<td>Chromium</td>
<td>2.05</td>
</tr>
<tr>
<td>Molybdenum</td>
<td>0.30</td>
</tr>
<tr>
<td>Tungsten</td>
<td>7.25</td>
</tr>
<tr>
<td>Aluminum</td>
<td>7.55</td>
</tr>
<tr>
<td>Titanium</td>
<td>1.30</td>
</tr>
<tr>
<td>Tantalum</td>
<td>9.15</td>
</tr>
</tbody>
</table>

Source: Ref 27

h/871 °C (1600 °F) or 48 h/850 °C (1562 °F)], gives CMSX-2 cubicall γ' with a mean size of 0.45 μm (18 μm.), which optimizes creep response (Ref 27, 29). Similar γ' morphology and size are obtained with a 4 h/1079 °C (1975 °F) AC postsolution treatment. The morphology of the γ' in CMSX-2 with this ONERA-type aging treatment is shown in Fig. 17(a), which should be compared to the conventional irregularly shaped γ' particles with a mean size of 0.3 μm (12 μm.) shown in Fig. 17(b). The particles shown in Fig. 17(b) result from a 5 h/982 °C (1800 °F) AC + 48 h/850 °C (1562 °F) aging (Tₐ). Specimens of the Tₐ type at 760 °C (1400 °F) deform in a homogeneous manner in the early stage of creep (Fig. 18). The homogeneous nature of the deformation leads to a rapid strain hardening of the material, causing a decrease in the creep rate. The Tₐ heat treatment, which produces smaller, irregularly shaped particles, favors inhomogeneous deformation within the specimen due to the precipitate shearing during the early stages of creep (Fig. 19). In this case, the amplitude of primary creep is high, and the strain hardening of the material is achieved at a much later stage, compared with that of the Tₐ-type heat-treated specimens.

During creep at high temperature, the γ' precipitates coarsen in the form of rafts perpendicular to the stress axis. The kinetics of raft formation depend on the testing temperature, among other factors. At 1050 °C (1922 °F) under a stress of 120 MPa (17.4 ksi), the rafts form within a few hours (Fig. 20). The rafts have a high aspect ratio in the Tₐ-type heat-treated specimens in which the cubicall γ' precipitates are already aligned. The lateral extension of the γ' phase in the form of rafts causes the specimen to creep at a much lower rate, compared with the creep rate of the material in which the γ' phase coalesces irregularly. The CMSX-2 and CMSX-3 alloys that show this type of rafted γ' morphology possess very long rupture lives at high temperatures. In these alloys, the misfit between the γ and γ' phases is found to be negative at high temperatures (Ref 30).

Work by ONERA and Ishikawajima-Harima Heavy Industries Company, Ltd. (IHI) shows some interesting effects of the crystal orientation and heat treatments on the creep behavior and strength of several single-

It has been reported by ONERA that a high-temperature aging heat treatment (Tₐ) [16 h/1050 °C (1922 °F) air cooled (AC)] following solution treatment, with subsequent intermediate temperature aging [20
Crystal superalloys (Ref 31). The salient features can be summarized as follows:

- At intermediate temperatures (760 to 849 °C, or 1400 to 1560 °F), the creep behavior of nickel-base single-crystal superalloys is extremely sensitive to crystal orientation and γ' precipitate size. For a γ' size in the range of 0.35 to 0.5 μm (14 to 20 μm), the highest creep strength is obtained near [001], while orientations near the [111]-{001} boundary of the standard stereographic triangle exhibit very short creep lives. When the γ' size decreases to 0.2 μm (8 μm), the longest creep lives are exhibited, in decreasing order, by the crystals oriented near [111], [001], and [110]. The anisotropy in creep between the [001] and [111] orientations can therefore be reduced by appropriate precipitation heat treatments. The creep strengths, however, remain poor near the [011] orientation.

- At high temperatures (982 to 1049 °C, or 1800 to 1920 °F), the creep behavior of the single-crystal superalloys is much less sensitive to crystal orientation and γ' size than it is at intermediate temperatures. The [001]-oriented single crystals develop a rafted γ' structure normal to the tensile stress axis, while the γ' precipitates coarsen irregularly in the [111] specimens.

**Fatigue.** An important property that must be considered when selecting single-crystal superalloys for turbine blade applications is fatigue strength. Single crystals of CMSX-2 have been cast both under low- and high-gradient conditions and then subjected to high-cycle fatigue tests in the repeated tension mode at 870 °C (1598 °F) (Ref 29); the results are reported in Fig. 21. The fatigue resistance of specimens cast under a very high temperature gradient (laboratory conditions) is much superior to that of material cast under industrial conditions, primarily because of the very small pore size (<10 μm, or 40 μm) inherent in the high-gradient specimens. The single crystals cast under industrial conditions have a more heterogeneous structure where the interdendritic spacing and the level of porosity vary along the length of the bar. Specimens corresponding to the beginning of solidification exhibit better fatigue resistance than those corresponding to the end of solidification. Some fatigue tests were also performed on specimens in which a rafted γ' morphology was developed prior to testing. It is interesting to note that the fatigue behavior is not significantly affected by the rafted γ' morphology (Ref 29).

Strain-controlled, fully reversed low-cycle fatigue tests performed at 760 °C (1400 °F) confirm the much better fatigue behavior of single crystals cast under a high gradient (Fig. 22). In this type of test, the higher the deviation from the [001] orientation, the shorter the fatigue life. It can be seen in Fig. 22 that for a total strain range of 1.25%, the fatigue life is decreased by an order of magnitude when the crystal orientation, relative to the [001], moves away from 6 to 22°. The decrease in fatigue life is a consequence of the increase in stress level through the increase of elastic modulus. Because the plastic strain component at 760 °C (1400 °F) is small, the results can be plotted as total stress versus the number of cycles to failure (Fig. 22). In Fig. 22, the effect of crystalline orientation on the fatigue life of the industrially processed single crystals is not apparent, and all the results of low-gradient single crystals can be represented by a single curve.
Fig. 15 CMSX-2 element homogenization effect. Source: Ref 21

Fig. 16 CMSX-4 (heat VF 719) SX test of flat specimen, 25 mm wide by 1.25 mm thick by 100 mm long (1 in. by 0.05 in. by 4 in.). Specimen was cast, 99% solutioned, and double aged. Micrographs taken from longitudinal orientation. (a) 90X. (b) 365X. (c) 905X. (d) 905X
An examination of fracture surfaces shows that the cracks are initiated at microporosity, which indicates that these defects (microporosity) are of primary importance in determining the fatigue life of CMSX-2. The size of microporosity in industrial single crystals can be as large as 50 to 80 μm (2000 to 3200 μm), but is rarely more than 10 μm (400 μm), in single crystals cast under very high temperature gradients (laboratory conditions). The adverse effect of microporosity is also confirmed by the results obtained after hot isostatic pressing [1000 bars/1315 °C (2400 °F)] CMSX-2 single-crystal bars solidified under the low-temperature gradient. The fatigue strength of single crystals cast under low gradients after hot isostatic pressing can be improved to that of the high-gradient-processed specimens (Fig. 22). The techniques developed to hot isostatic press CMSX-2 ensure that no recrystallization occurs internally within the test bars during the hot isostatic press cycle.

**Oxidation and Hot Corrosion.** Coatings perform well with single-crystal alloys, particularly when they are optimized to the base alloy system, as shown by the work with PWA 1480 alloy in Fig. 23. The absence of grain boundaries, rosette clusters of carbides, and elemental segregation in heat-treated single-crystal alloys contribute to improved coating performance.

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Fig. 21 Effect of thermal gradient and heat treatments on the high-cycle fatigue behavior of CMSX-2 at 870 °C (1598 °F) with frequency of 50 Hz. Source: Ref 29

Fig. 22 Effect of thermal gradient, orientation, and hot isostatic pressing on the strain-controlled low-cycle fatigue behavior of CMSX-2 (fully reversed, with frequency of 0.33 Hz) at 760 °C (1400 °F). Numbers represent the deviation, in degrees, from the [001] orientation. (a) Strain versus cycles to failure. (b) Stress versus cycles to failure. Source: Ref 29
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