

# Materials for Gas Turbines

by R. J. E. Glenny, J. E. Northwood, and A. Burwood-Smith

The major components of gas turbines and their materials requirements are outlined. The properties, advantages, and limitations of materials which have been and are being used in gas turbines are then described in the following groups: aluminium alloys, magnesium alloys, titanium and its alloys, steels, and finally nickel- and cobalt-base alloys. The applications of the last group of alloys and the importance of their metallurgical stability in turbine blades, discs, nozzle guide vanes, and sheet metal components are discussed. Corrosion and its prevention are then considered, followed by sections on experimental materials including alloys of chromium and the refractory metals, ceramics, and fibre composites. Future trends and prospects are discussed briefly in a concluding section.

Further progress in the development of aero gas turbines depends on the ability of materials engineers and scientists to provide materials with improved strength/weight ratios and with better retention of their strength properties at higher temperatures. Additionally, any new material must possess adequate resistance to fatigue, corrosion, and mechanical shock. For industrial gas turbines, weight and size considerations tend to be less significant, but corrosion resistance and metallurgical stability assume greater importance.

The applicability of a material depends on the engine design – whether it is a subsonic or supersonic turbojet, a lift engine, or a shaft-power engine – and the engine duty, e.g. whether civil or military aircraft, industrial, or marine power. Sophistication in British aero-engine design has progressed from single-shaft engines of the Avon<sup>1</sup> type (Fig.1) through the two-shaft Conway<sup>2</sup> (Fig.2) and Olympus<sup>3,4</sup> types to the recently developed three-shaft RB 211<sup>5</sup> engines (Fig.3) involving high air mass flows and necessitating large but light compressors. The use of higher turbine inlet temperatures has been made possible by air cooling the high pressure (HP) turbine blades and nozzle guide vanes. Small shaft-power engines with uncooled turbines are used in helicopters and hovercraft (e.g. Gnome) and for automotive propulsion (e.g. Rover), the latter requiring the use of heat exchangers<sup>6</sup> in order to obtain acceptable fuel consumption.

In the highly competitive aero-engine market, material

selection is strongly governed by technical and economic requirements. Since simultaneous improvement in the major engine parameters is often impossible, some compromise is essential. A high-temperature, low-weight engine demands improved blade cooling and more advanced, and possibly more expensive, materials in both compressor and turbine. Good economy of operation implies refined mechanical design which puts greater demands on all materials. On the other hand, a rugged, reliable, low-cost engine involves the use of well-proven, robust, and relatively cheap materials. Processing and fabrication procedures influence not only property levels but the cost of material and component. In classing a material as 'technically' good, it is increasingly implied that it is economically attractive for the finished component.

This review is concerned with recent developments in materials and processes for gas turbines and in particular with the merits and limitations of both potentially useful and established materials for the critical components of UK aero gas turbines.

## Gas Turbine Components and Materials Requirements

The properties and characteristics required for gas turbine materials are strongly associated with the individual components in the compressor, combustion, turbine, and exhaust stages of the engine cycle.

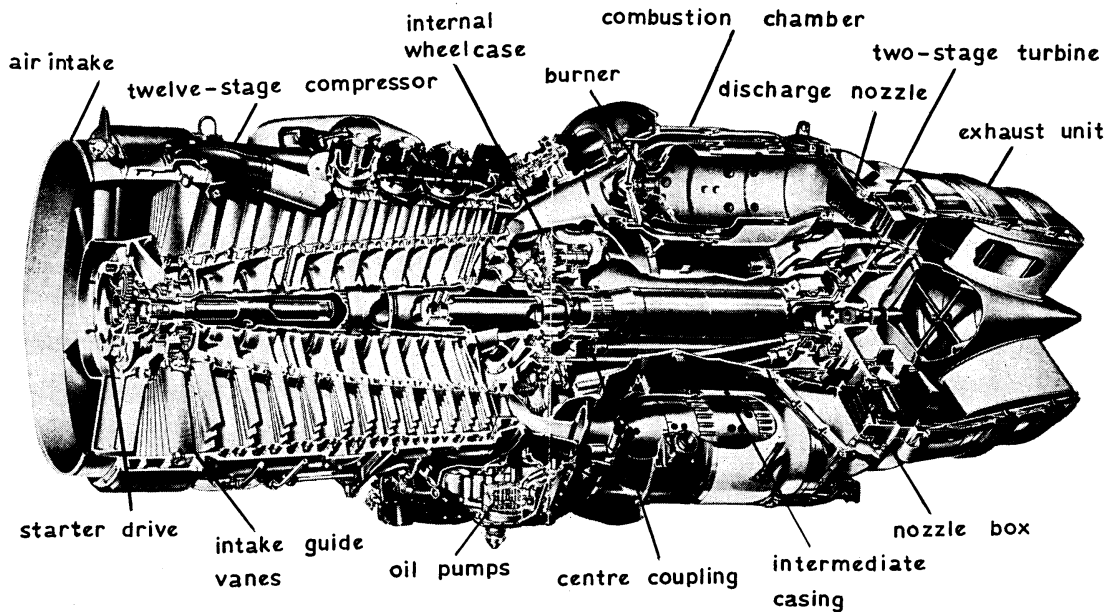
Compression of the air before combustion and expansion through the turbine is effected by a centrifugal or axial compressor, the latter being used for high-performance engines involving high-pressure ratios and high air mass flows. As shown in Fig.2, the axial flow compressor consists of a twin-spool rotor assembly carrying several rows of rotating blades, each of which is followed by a row of stator blades mounted in a casing. The rotating parts (discs, blades, shafts, and bearings) are more highly loaded than stationary components such as the stators, inlet guide vanes, and casings, and have therefore merited more attention with respect to materials research and development. In the largest engines, compressor temperatures may range from  $-50^{\circ}\text{C}$  at the inlet to  $+500^{\circ}\text{C}$  at delivery.

Selection of a compressor blade material depends largely on its specific tensile and stiffness properties and its fatigue and creep characteristics, but corrosion resistance must be adequate and, for aero-engines, resistance to erosion and impact damage by 'foreign bodies' such as sand, stones, or birds is imperative. Titanium alloys and steels are widely used in the more powerful engines, except in the hottest stages, which require nickel-base alloys. For the blading of small shaft-power engines, 12–14%Cr martensitic steels have been favoured. In lift engines, where low weight is all-important, glass-fibre reinforced plastics have been applied.<sup>7</sup>

The compressor (and turbine) discs operate at high rotational speed, and great care must therefore be taken in design and material selection to avoid the catastrophic results of failure and disintegration in service. Steels and titanium- and nickel-base alloys are used according to the service temperature. Disc

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1 Cutaway view of Avon engine

[Courtesy of Rolls-Royce (1971) Ltd]

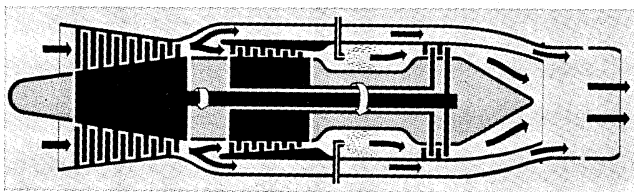
alloys are initially selected on the basis of uniaxial tensile and creep properties. Limited ductility is needed for stress redistribution at root fixings and bolt holes during service. The crack tolerance of materials, evaluated by cyclic spinning tests on model and full-size discs as well as by laboratory crack propagation and fracture toughness tests, increasingly affects their selection. In addition to fatigue strength at  $10^4$ – $10^5$  cycles, a knowledge of the number of cycles required to initiate a measurable crack and of its rate of growth in the disc is essential. Fatigue behaviour is more likely to be improved by control of microstructure than by alloy development.

Combustion must be accomplished with the minimum loss of pressure, and with the maximum heat release for the limited space available and the limited temperature which the turbine materials can withstand. It is carried out in tubular (multiple) chambers or annular (single) chambers, each with an inner and outer casing fabricated from heat- and creep-resistant alloys that must withstand corrosion by the combustion products, fatigue due to vibrational stresses, and creep due to thermal stresses, and must also be easily formed and joined. Air flow is delivered from the compressor to the annulus and into the chamber via apertures in the wall or slots in rings that stiffen and strengthen the inner shell at intervals. Thus the chamber wall is film- or effusion-cooled to an average temperature of about  $800^\circ\text{C}$ , although local temperatures may reach  $1000^\circ\text{C}$ . The life of such structures is largely determined by the properties of welded joints and of adjacent heat-affected zones.

Material selection is also influenced by the resistance of the welded structure to embrittlement in service and by its facility for repair by welding.

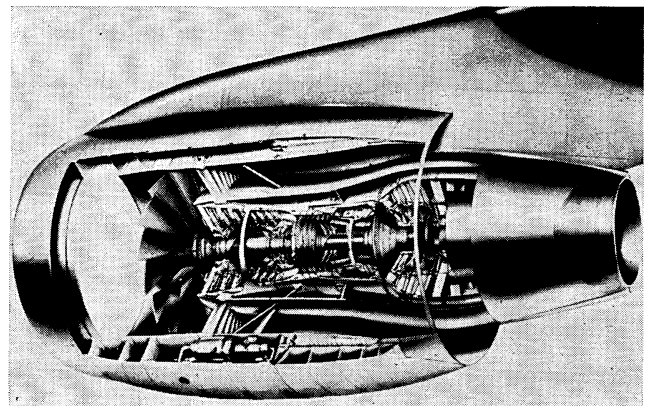
The turbine consists of two or more stages, each containing a row of stationary nozzle guide vanes and a row of rotating blades. Except for small shaft-power aero-engines and industrial gas turbines, the HP stage is cooled, so that the mean metal temperatures may be  $200^\circ$ – $300^\circ\text{C}$  lower than the gas temperature. The HP nozzle guide vanes invariably operate  $50^\circ$ – $100^\circ\text{C}$  higher than the HP rotating blades and, though they experience greater corrosion and higher transient thermal stresses, the only mechanical loads are due to gas bending. Although protective coatings and internal cooling have reduced the corrosion and thermal shock problems respectively of cast nickel- or cobalt-base alloy vanes, these problems recur in more advanced (i.e. higher-temperature) engines.

The HP turbine rotor blade probably operates under more rigorous conditions of stress and temperature than any other



2 Schematic diagram of Conway engine

[Courtesy of Rolls-Royce (1971) Ltd]



3 Cutaway view of RB 211

Courtesy of Rolls-Royce (1971) Ltd

component. Apart from the need to withstand centrifugal tensile loads, a satisfactory blade material must be resistant to superimposed fatigue stresses (which are vibrational, mechanical, and thermal in origin), to impact, and to corrosion and erosion by hot oxidizing gases. Invariably, nickel-base alloys are used. Higher turbine entry temperatures (TET) have been achieved by internal cooling, and so the merits of new materials must be considered in terms of the TET that can be achieved. Because the centrifugal loading is a function of blade density, materials are compared in terms of specific tensile and creep strength; furthermore, their density must not be excessive lest a weight penalty should arise not only from the heavier blades but the heavier disc that would be needed to support them. As higher values of TET are required to achieve greater thrust or power output, there is a continuous effort to find improved materials for cooled and uncooled turbine blading.

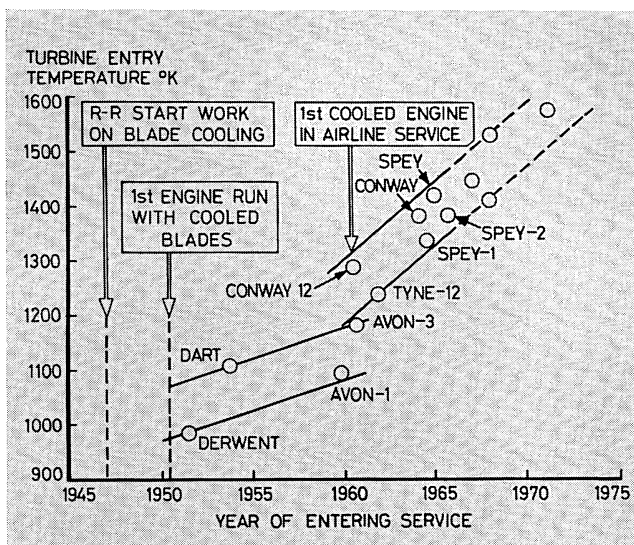
The exhaust system, generally comprising thrust reversers, noise suppressors, and variable propelling nozzles, discharges the turbine gases at 550°–850°C to the atmosphere at velocities up to 365 m/s. Outer casings are of heat-resistant steels and jet pipe liners are of nickel-base alloys. Afterburning or reheat, used to produce additional thrust for short periods, creates flame temperatures as high as 1700°C and, as in the turbine and combustion areas, involves interaction between materials and cooling technologies.

Air cooling is an important design feature of the gas turbine engine. Selection of suitable materials for the combustion and turbine areas is inevitably associated with the extent to which cooling by the working fluid, air, is possible. A controlled amount of air from the compressor is circulated around components in order to prevent excessive absorption of heat from the gas stream and thereby to ensure their safe operation. It is not generally appreciated<sup>8</sup> that about one third of the air delivered from the compressor is used to cool the flame tube walls. Air bled from the LP compressor pressurizes the oil seals and cools the lubricating oils. High-pressure air externally cools the faces of the turbine discs and internally cools the nozzle guide vanes and turbine blades, thus allowing the gas temperature to be raised while the actual blade temperature

remains at a value low enough to preserve the material properties.

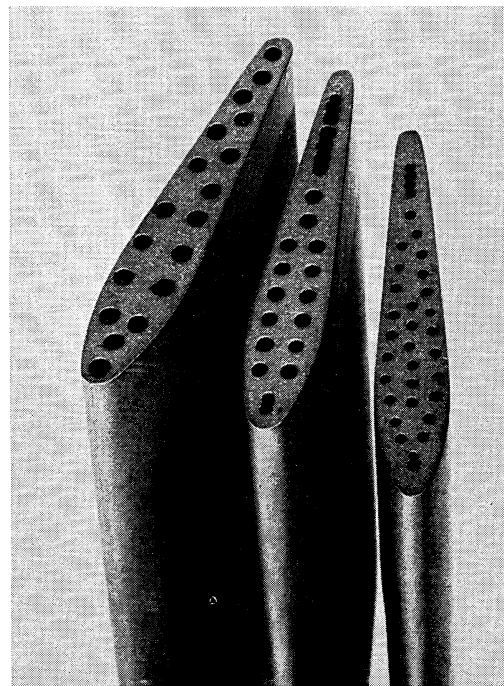
British production engines have employed<sup>9, 10</sup> cooled blades since 1960, their introduction having resulted in an immediate increase in TET of about 100°C (see Fig.4). Techniques for making convectively cooled cast blades, involving the use of fused silica tube cores subsequently leached from the casting (see Fig.5) were developed at the National Gas Turbine Establishment as early as 1949. Similar techniques were subsequently employed for making cooled HP turbine blades for Olympus engines, particularly for the Olympus 593. Progressive improvements in both cooling technology and materials have subsequently resulted<sup>11</sup> in an average increase in TET of about 25°C per year. In advanced engines, such as the RB 211, the gas temperature is close to the melting point of the blade material, but the mean metal temperature is about 300°C lower. Gresham<sup>11</sup> has stated that in 1969 the rate of improvement due to cooling was approximately three or four times that obtained through materials development. However, for a blade cooling efficiency (i.e.  $\frac{T_G - T_B}{T_G - T_C}$  where  $T_G$ ,  $T_B$ , and  $T_C$  are respectively the mean gas, blade, and inlet cooling air temperatures) of, say, 0.5, any increase in  $T_B$  resulting from alloy development produces a two-fold increase in  $T_G$ .

Further gains by cooling alone depend on the efficient utilization of the cooling air rather than an increase in quantity which would lead to performance losses. Because of the complex internal geometry of cooled blades and of the high transient and steady thermal stresses (which can be several times that of the steady centrifugal stress), failure by creep can occur in much shorter times than that predicted from laboratory constant-load creep tests. In order to increase blade life, studies of the effects of creep behaviour of thin sections and stress concentrations associated with the cooling passage design, and of complex stress redistribution are being



4 Effect of advances in cooling technology on turbine entry temperatures of UK engines

[Courtesy of Rolls-Royce (1971) Ltd



5 Aerofoil shapes with cooling passages, cast by NGTE in 1949

TABLE I  
Nominal Chemical Compositions and Densities of Light Alloys

| Alloy                   | Composition, wt-% |          |      |      |          |     |           |          |      |     |      |                                  | Density,<br>10 <sup>3</sup> kg/m <sup>3</sup> |
|-------------------------|-------------------|----------|------|------|----------|-----|-----------|----------|------|-----|------|----------------------------------|-----------------------------------------------|
|                         | Si                | Fe       | Ti   | Al   | Ni       | Zr  | Sn        | Zn       | Mg   | Cu  | Be   | Mn                               |                                               |
| ZW 6                    | —                 | —        | —    | —    | —        | 0.7 | —         | 5.5      | Bal. | —   | —    | —                                | 1.82                                          |
| RZ 5                    | —                 | —        | —    | —    | —        | 0.7 | —         | 4.0      | Bal. | —   | —    | —                                | 1.81                                          |
| ZT 1                    | —                 | —        | —    | —    | —        | 0.7 | —         | 2.2      | Bal. | —   | —    | —                                | 1.83                                          |
| MSR-B                   | —                 | —        | —    | —    | —        | 0.7 | —         | —        | Bal. | —   | —    | —                                | 1.82                                          |
| ZRE-1                   | —                 | —        | —    | —    | —        | 0.7 | —         | 2.2      | Bal. | —   | —    | —                                | 1.80                                          |
| ZE 63A                  | —                 | —        | —    | —    | —        | 0.6 | —         | 5.8      | Bal. | —   | —    | —                                | 1.87                                          |
| Be I-400                | —                 | —        | —    | —    | —        | —   | —         | —        | Bal. | —   | Bal. | —                                | 1.85                                          |
| RR 50                   | 2.2               | 1.1      | 0.2  | Bal. | 1.3      | —   | 0.05 max. | 0.1 max. | 0.1  | 1.4 | —    | 0.1 max.                         | 2.75                                          |
| RR 250                  | 0.3               | 0.5 max. | 0.2  | Bal. | 1.0      | —   | —         | —        | —    | 5.0 | —    | 0.25                             | 2.84                                          |
| RR 350                  | 0.3 max.          | 0.5 max. | 0.2  | Bal. | 1.6      | 0.2 | —         | —        | —    | 5.0 | —    | 0.25                             | 2.84                                          |
| Hiduminium<br>54        | 0.15              | 0.3 max. | 0.13 | Bal. | 0.1 max. | —   | —         | 0.1 max. | 0.2  | 6.0 | —    | 0.25                             | 2.84                                          |
| Hiduminium<br>RR 58     | —                 | 1.0      | 0.1  | Bal. | 1.2      | —   | —         | —        | 1.5  | 2.5 | —    | —                                | 2.75                                          |
| Hiduminium<br>100 (SAP) | —                 | —        | —    | Bal. | —        | —   | —         | —        | —    | —   | —    | —                                | 2.70                                          |
|                         |                   |          |      |      |          |     |           |          |      |     |      | 13Al <sub>2</sub> O <sub>3</sub> | 2.70                                          |

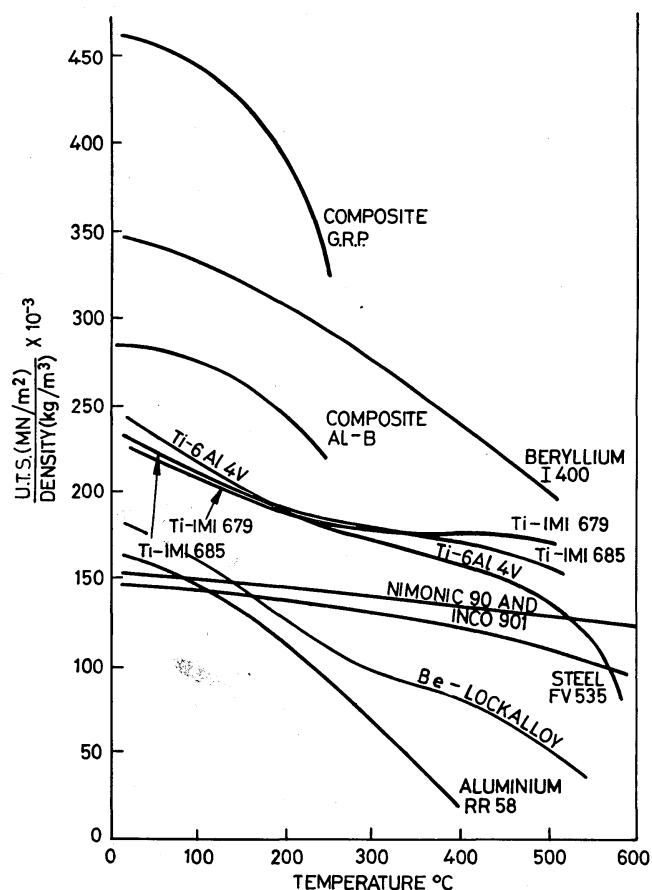
carried out parallel with efforts to reduce fabricating costs. In 1969, the cost per ounce of a cooled blade approached<sup>10</sup> that of the current official price of gold. More sophisticated designs of cooled blade, such as the use of transpiration cooling<sup>12</sup>

involving porous walls, are being considered in the same light as exotic materials, first cost and blade life being as important as the performance gain.

### Light Alloys

#### Aluminium Alloys

Forged aluminium alloys such as Hiduminium RR 58\* (see Table I) were used<sup>13-15</sup> for the centrifugal impellers of the early Whittle and Derwent engines, and retained for similar use in gas turbines for road propulsion vehicles.<sup>16</sup> They are useful in gas turbines up to about 180°C (see Fig. 6), which is close to the limit of usefulness for rotating components of precipitation-hardened systems so far developed. With the development of the axial compressor, higher-strength materials of superior erosion and fatigue resistance and fracture energy were needed for the blading, and so aluminium alloys were largely displaced first by steels and then by titanium alloys. Nevertheless, some compressor stages of Rolls-Royce engines



6 Specific tensile strength v. temperature for compressor blade materials

TABLE II  
Melting Points and Densities of Base Metals

| Metal           | Melting point, °C   | Density, 10 <sup>3</sup> kg/m <sup>3</sup> |
|-----------------|---------------------|--------------------------------------------|
| Magnesium       | 650                 | 1.74                                       |
| Aluminium       | 660                 | 2.70                                       |
| Titanium        | 1668                | 4.5                                        |
| Beryllium       | 1277                | 1.84                                       |
| Boron           | c. 2030             | 2.5                                        |
| Carbon          | 3730                | 2.25                                       |
| Iron            | 1536                | 7.85                                       |
| Nickel          | 1455                | 8.9                                        |
| Cobalt          | 1492                | 8.71                                       |
| Chromium        | 1850                | 7.2                                        |
| Niobium         | 2468                | 8.55                                       |
| Molybdenum      | 2610                | 10.2                                       |
| Tantalum        | 2996                | 16.6                                       |
| Tungsten        | 3410                | 19.3                                       |
| Silicon nitride | Sublimes at 1900    | 3.0                                        |
| Silicon carbide | Dissociates at 2700 | 3.1                                        |

\*Reference to alloys in the text is invariably made to their trade marks or trade names, the originators of which are given in the Appendix. The nominal compositions of commercial and experimental alloys are given in the relevant tables.

such as the Spey and the Avon are in aluminium alloy. However, because of their low density (*see* Table II), low cost, and ease of castability, cast aluminium alloys, e.g. RR 250 and RR 350,<sup>17</sup> are used in compressor casings, where they are competitive with steels and magnesium alloys. Hiduminium 54 is used in the RB 211 engine as a fan-disc shroud serving to redirect warm air for de-icing purposes. In lift engines where high thrust per unit weight is the major requirement, the first compressor stage may be in forged aluminium alloy. Sintered aluminium powder (SAP)<sup>18, 19</sup> shows better retention of strength at temperature than RR 58, but because of its relatively high cost, low cold strength, and generally low ductility it was not applicable to gas turbines, and is no longer commercially available in the UK. Fibre composites with an aluminium matrix may be competitive with titanium alloys for compressor blading if they can offer gains in specific strength and specific modulus in association with reduced weight and adequate corrosion and impact resistance (*see* below).

### Magnesium Alloys

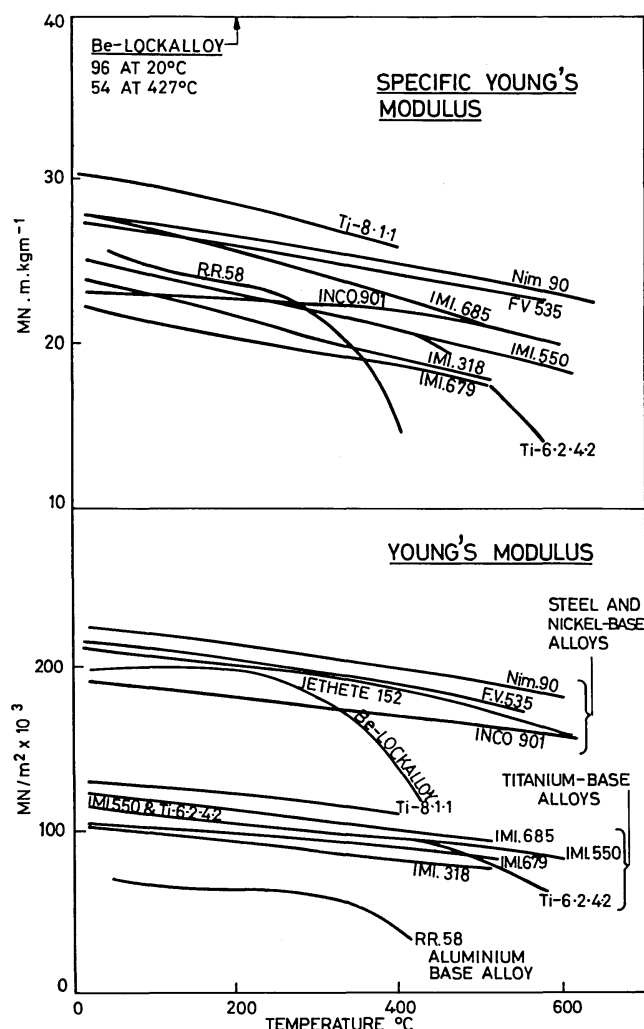
The density of magnesium alloys is about two-thirds that of aluminium alloys and less than a quarter that of steels, and so they are attractive for low-weight structures. Although unprotected magnesium is less affected than mild steel when exposed to normal atmospheres and is free from intergranular attack, it is anodic to most other metals; so protection is essential, particularly in marine environments. Chromate or anodic coatings, followed by waterproofing with a surface sealing resin overpainted with an epoxy primer and polyurethane finish, have been developed.<sup>20, 21</sup> However, at joints such as those at compressor half-casings, wet assembly is needed for initial sealing and at overhaul. Mechanical conversion coatings are preferred where high coating integrity is imperative, as it is, for example, for internal surfaces of gear boxes.

Although specific strength of wrought Mg-Zr-Zn alloys such as ZW6 (*see* Table I) compares favourably with that of aluminium alloys, their greater notch sensitivity in tension, low fatigue/ultimate strength ratio, and low Young's modulus (typically 44 GN/m<sup>2</sup>) have prevented their use for blade applications. Ignoring cost considerations, improvements in stiffness by, say, fibre reinforcement, or in strength and ductility by controlled solidification would not alleviate corrosion problems. The high-strength (180–240 MN/m<sup>2</sup> at 20°C) cast alloys are of the Mg-Zn-Zr and Mg-Ag-Zr type, with thorium or rare-earth additions for further strengthening (*see* Table I). The specific stiffness and strength up to 260°C of premium quality castings in a MSR-B alloy are superior<sup>20</sup> to those of high-strength cast aluminium alloys. ZT 1 alloy has particularly good castability and retains its strength to higher temperatures than MSR alloy. The properties of ZE 63 alloy<sup>22, 23</sup> are achieved by heat treatment in hydrogen, resulting in a uniform dispersion of the hydrides of zirconium and rare earths in sections up to several millimetres thick, and releasing the zinc (in the zinc and rare-earth compounds) for strengthening purposes. This alloy has been specified<sup>23</sup> for cast thrust reversers in the RB 211 engine. Cast magnesium alloys have been used<sup>24</sup> for air intakes (RZ 5 alloy in Gyron Junior) and compressor casings<sup>25, 26</sup> (ZT 1 alloy in Gazelle, MSR-B alloy in Conway Co-12, and RZ 5 alloy in the M45H engine). However, the casings of the larger engines require greater stiffness which can often be achieved without a weight penalty by the use of pressed steel. In engines for vertical take-off or short take-off aircraft, weight saving is particularly important; the Pegasus engine for the Harrier contains<sup>27</sup> about 90 kg of magnesium. Although the cost of protection is a

consideration, closer cooperation between designer and metallurgist could conceivably lead to greater usage of magnesium alloys.

### Beryllium

The low density of beryllium – about two-thirds that of aluminium and about two-fifths that of titanium (*see* Table II) – and its high Young's modulus (about 300 GN/m<sup>2</sup>) are attractive for lighter and more compact power plant.<sup>28</sup> Practically all beryllium is a powder metallurgical product. The properties of commercial grades vary markedly with purity (e.g. oxide content), thermomechanical working, and grain size. Because of its close-packed hexagonal structure with an inherently low axial ratio and with the basal plane the predominant slip plane, anisotropy of properties,<sup>29</sup> particularly ductility, is evident. Lack of ductility is marked in the short-transverse direction, and the resulting poor resistance to impact, typified by the ease of fragmentation of wedge and plate specimens in 1400 grade<sup>30</sup> (*see* Table I and Fig.6) has been mainly responsible for its exclusion from aero-engine compressors. Despite intensive efforts involving controlled changes of composition, triaxial forging (to induce isotropy of properties), and



7 Modulus and specific modulus v. temperature for compressor blade materials

[Courtesy of Rolls-Royce (1971) Ltd

TABLE III

\*Al alloys only  
RR engines; Ref. 72  
GE and USA engines; Ref. 34

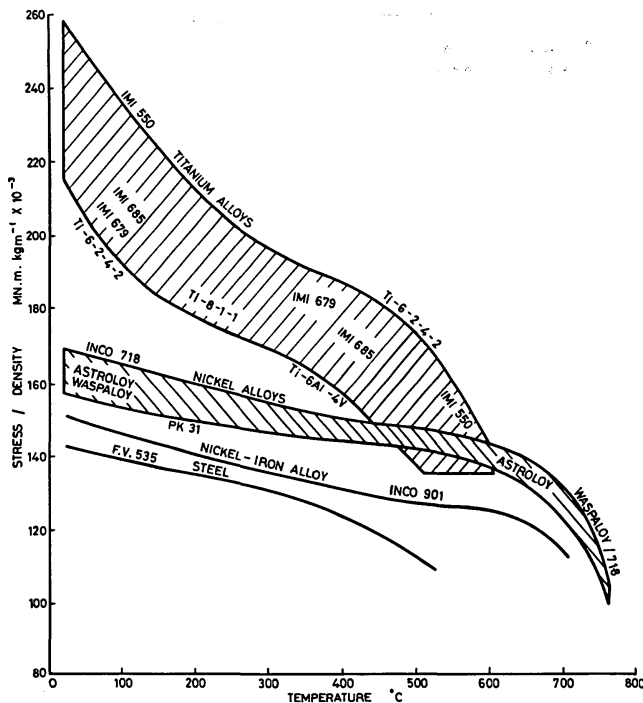
## Titanium and its Alloys

corrosion resistance and adequate strength at temperature (up to 550°C for the more complex alloys). The aluminium alloys and to an appreciable extent the steels used in the earlier engines have been largely replaced by titanium alloys (see Table IV and Figs.6-8) which are now extensively applied in both the LP, (fan) and IP compressor stages. The use of titanium alloys instead of steels for compressor blades permits the use of titanium discs of reduced section compared to that of steel and so an all-titanium blade-disc assembly shows a greater weight saving (about 52%) than would be expected (about 42%) from the difference in the densities of the respective materials.

The considerable research and development into titanium and its alloys is reflected in the Proceedings of the First (1968) and Second (1972) International Conferences, the latter involving four volumes and over 2700 pages. Yet, as an engineering material, titanium is only some thirty years old. The philosophy of titanium alloy development has been much the same in the UK and the USA, although the latter tend to use more sophisticated heat treatments. Composition and processing have been rationally varied to produce microstructures that will permit a desirable trade-off in properties. Alpha alloys are used because of their excellent low-temperature ductility, high creep strength, and good weldability. The chief substitutional solid solution elements are aluminium, tin, and zirconium, although carbon and oxygen are important interstitial strengtheners. In the UK, Ti-2Cu (IMI 230)<sup>38</sup> possesses the formability and weldability of unalloyed titanium and has

TABLE IV

| Alloy           | Composition, wt.-% |           | Ti   | Al   | Mo  | Zr  | Sn  | Cu  | V   | Others | Density,<br>10 <sup>3</sup> kg/m <sup>3</sup> |
|-----------------|--------------------|-----------|------|------|-----|-----|-----|-----|-----|--------|-----------------------------------------------|
|                 | Si                 | Fe        |      |      |     |     |     |     |     |        |                                               |
| IMI 230         | —                  | —         | Bal. | —    | —   | —   | —   | 2·5 | —   | —      | 4·56                                          |
| IMI 317         | —                  | —         | Bal. | 5    | —   | —   | 2·5 | —   | —   | —      | 4·46                                          |
| IMI 318A        | —                  | —         | Bal. | 6    | —   | —   | —   | —   | 4   | —      | 4·42                                          |
| IMI 550         | 0·5                | —         | Bal. | 4    | 4   | —   | 2·0 | —   | —   | —      | 4·60                                          |
| IMI 679         | 0·2                | —         | Bal. | 2·25 | 1   | 5   | 11  | —   | —   | —      | 4·84                                          |
| IMI 684         | 0·3                | —         | Bal. | 6    | —   | 5   | —   | —   | —   | IW     | 4·48                                          |
| IMI 685         | 0·25               | —         | Bal. | 6    | 0·5 | 5   | —   | —   | —   | —      | 4·45                                          |
| Ti-811          | —                  | 0·3 max.  | Bal. | 8    | 1   | —   | —   | —   | 1   | —      | 4·37                                          |
| TMCA<br>Ti-6242 | —                  | 0·25 max. | Bal. | 6    | 2   | 4   | 2   | —   | —   | —      | 4·45                                          |
| Ti-662          | —                  | 0·7       | Bal. | 5·5  | —   | —   | 2   | 0·7 | 5·5 | —      | 4·58                                          |
| TMCA<br>Ti-6246 | —                  | 0·15 max. | Bal. | 6    | 6   | 4   | 2   | —   | —   | —      | 4·68                                          |
| TMCA<br>Ti-11   | 0·1                | —         | Bal. | 6    | 1   | 1·5 | 2   | —   | —   | 0·35Bi | 4·48                                          |
| GE<br>Ti-17     | —                  | —         | —    | 5    | 4   | 2   | 2   | —   | —   | 4Cr    | —                                             |



8 Specific tensile strength v. temperature for compressor and turbine disc materials

been used for the casings of the Olympus 593 and the ducting of Spey and Conway engines. Of the simple  $\alpha$  alloys, Hubbard<sup>39</sup> states that only the commercially pure grades and the 5Al-2.5Sn alloy have withstood the test of time.

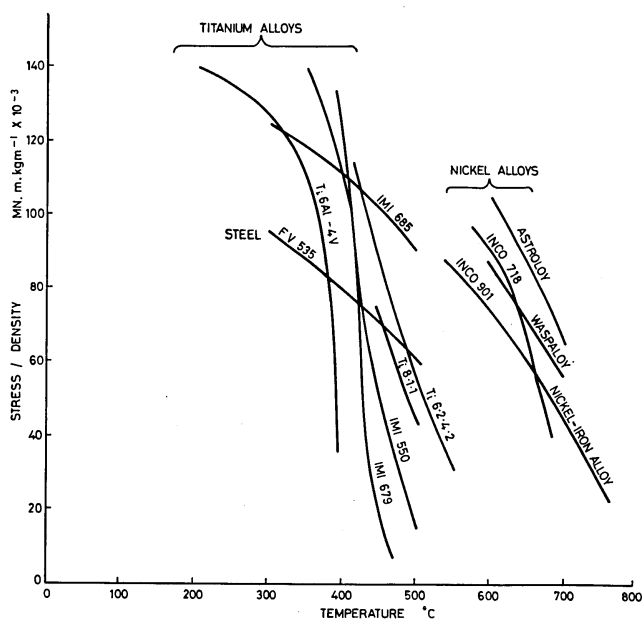
The strength limitations of the simple  $\alpha$  alloys are reduced by the addition of  $\beta$ -stabilizers such as chromium, manganese, iron, molybdenum, vanadium, and silicon. The  $\alpha + \beta$  alloys, heat-treated in the  $\alpha + \beta$  field (see Table V) include Ti-6Al-4V<sup>40</sup> (TMCA 6-4 and IMI 318A), IMI 550 (formerly Hylite 50<sup>41</sup>), TMCA Ti-6Al-2Sn-4Zr-6Mo alloy, and GE Ti-17,<sup>34</sup> which have 4-6% of  $\beta$ -forming additions (molybdenum or vanadium). Such alloys show good creep resistance up to about 400°C. The most common  $\alpha + \beta$  alloy, Ti-6Al-4V, has a fine equiaxed structure of primary  $\alpha$  and transformed  $\beta$ . Its specific strength and stiffness characteristics are shown in Figs.6 and 7. It accounts for about 50% of all titanium metal production and is widely used in the fan and LP compressor stages of engines.  $\alpha + \beta$  alloys of improved creep strength and with silicon as a

strengtheners have been developed in the UK by Jessop Saville Ltd and IMI. IMI 550 (see Table IV) is widely used<sup>41</sup> in UK engines, e.g. Olympus 593 and Pegasus, for compressor blades and discs (see Figs.7 and 8). For use at temperatures up to 450°C, the  $\beta$ -stabilizing additions have been reduced and the  $\alpha$  additions optimized in commercial alloys such as IMI 679 and TMCA Ti-6Al-2Sn-4Zr-2Mo alloy which are heat-treated in the  $\alpha + \beta$  field. IMI 679 has been used<sup>42, 43</sup> for compressor discs in UK engines and for the HP turbine disc of the Rolls-Royce RB 162 lift engine. Even greater creep resistance has been achieved by developing near  $\alpha$  alloys for heat treatment in the  $\beta$  field, e.g. IMI 684 and 685,<sup>44</sup> and TMCA Ti-11.<sup>45</sup> Alloys such as IMI 685 have superior hot strength and fracture toughness, but inferior ductility (see Table V) and low-cycle fatigue (LCF) resistance to  $\alpha + \beta$  alloys such as Ti-6Al-4V, and are capable of operating for prolonged times at 500°C. Although the creep of titanium alloys at low strain rates ( $< 5 \times 10^{-4}$ %/h) is controlled<sup>46</sup> by the drag of solute atoms induced by moving dislocations, in molybdenum-containing alloys such as IMI 685 eutectoid precipitation on  $\alpha$ -plate boundaries controls creep deformation. This alloy was selected for extensive use in the RB 211 engine because of its combination of high creep strength (see Fig.9), fracture toughness, and LCF resistance, and good weldability. The seven-element alloy, Ti-11, was developed<sup>45</sup> by Titanium Metals Corporation as a 'super- $\alpha$ ' alloy for service up to 550°C. This alloy is about 170°C better than Ti-6Al-4V. Limited information is available<sup>47, 48</sup> on Russian alloys; 100 h rupture strength at 600°C of 250, 162, and 323 MN/m<sup>2</sup> respectively have been reported for alloys ST 1 (Ti-Al-Zr-Sn), ST 3 (Ti-Zr-Al-Mo-Fe), and ST 4 (Ti-Zr-Sn-Mo).

Creep strength is often the limiting parameter in advanced compressor design, but it must be accompanied by adequate elevated temperature tensile properties, LCF resistance, and both metallurgical and surface stability. Below a certain temperature, which varies with the alloy, LCF strength rather than creep strength is the limiting criterion for safe design. Under the combined effect of time and temperature, the most creep-resistant alloys tend to embrittle, particularly those of high aluminium content and with the eutectoid precipitates in the structure. Although there is greater alloying scope with  $\beta$ -forming additives, the higher density of  $\beta$  alloys and their reduced oxidation resistance, leading to surface embrittlement in service above 500°-550°C, will limit their usefulness unless protective coatings are developed and accepted, as in the case of nickel-base alloys (see below). Russian work indicates<sup>49</sup> that the creep of binary alloys is much accelerated in high-velocity gas streams. Substructure stabilization and texture strengthening by thermal mechanical processing merit greater attention in creep-resistant alloy development.

TABLE V  
Tensile Properties of Representative UK Titanium Alloys (5 cm dia. bar)

| Alloy    | Type             | Heat-treated in field | Test temp., °C | 0.2% PS, MN/m <sup>2</sup> | UTS, MN/m <sup>2</sup> | Elongation at fracture, % | Reduction of area, % |
|----------|------------------|-----------------------|----------------|----------------------------|------------------------|---------------------------|----------------------|
| IMI 318A | $\alpha + \beta$ | $\alpha + \beta$      | 20             | 930                        | 1020                   | 14                        | 38                   |
|          |                  |                       | 400            | 580                        | 700                    | 19                        | 65                   |
|          |                  |                       | 500            | 440                        | 600                    | 27                        | 72                   |
| IMI 679  | $\alpha$         | $\alpha + \beta$      | 20             | 1120                       | 1160                   | 15                        | 40                   |
|          |                  |                       | 400            | 660                        | 860                    | 17                        | 48                   |
|          |                  |                       | 500            | 620                        | 800                    | 19                        | 48                   |
| IMI 685  | $\alpha$         | $\beta$               | 20             | 900                        | 1020                   | 11                        | 21                   |
|          |                  |                       | 400            | 600                        | 760                    | 18                        | 30                   |
|          |                  |                       | 500            | 520                        | 700                    | 20                        | 36                   |



9 Specific 0.1% creep strain (100 h) v. temperature for compressor and turbine disc materials

In recent years, increased sophistication in alloy development has been accompanied by an increase in the number of design and operational requirements, and by greater attention to 'metallurgical synthesis',<sup>50</sup> i.e. the creation of the desired microstructure by control of composition and processing, particularly when applied to components of large size and complex geometry such as fan discs. Forging in the  $\beta$  field facilitates deformation and gives<sup>51</sup> a more homogeneous structure but, without good process control, incurs the risk of increased oxygen/hydrogen pick-up and excessive grain growth. Fracture toughness, notched tensile strength, and creep resistance are often improved by  $\beta$ -forging, but tensile ductility is generally lower than for  $\alpha$ - $\beta$  alloys. For weakly  $\beta$ -stabilized alloys,  $\beta$ -forging has been considered acceptable. Similar disadvantages and reservations apply to heat treatment in the  $\beta$  field. Gatorizing<sup>52</sup> (i.e. a process of hot isothermal die forging such that the alloy is worked in a temporarily superplastic condition) has been recently advocated for making Ti-6Al-4V compressor blades in order to permit reduced input weight and greater complexity of shape.

The susceptibility of titanium alloys to stress corrosion cracking (SCC) is well established. Their behaviour in aqueous chlorides has been reviewed by Brettell.<sup>53</sup> Sea water, fuming nitric acid, methanol-halide-water systems, organic halogen derivatives, and the metals cadmium, silver, and mercury have given rise to room or elevated temperature cracking. Stringent shop floor precautions are necessary during fabrication and heat treatment to exclude halogen-containing materials such as trichloroethylene. In aqueous solutions it is not known whether the mechanism is one of anodic dissolution or of hydrogen embrittlement. In hot salt environments (above about 260°C), exposure of titanium alloys under stress can result<sup>53, 54</sup> in embrittlement and much reduced creep and fatigue resistance. Rideout *et al.*<sup>55</sup> and Gray<sup>43</sup> considered that hydrogen produced by corrosion (i.e. of the metal by acid released during hydrolysis of the halide) embrittles the metal and causes initiation of cracks that grow until they are

blunted by plastic deformation in the more ductile interior; further crack growth is controlled by hydrogen diffusion. Although the dependence of SCC susceptibility on alloy composition is often obscured by structural variables, it is apparent that  $\alpha$ -stabilizers such as aluminium increase susceptibility while isomorphous  $\beta$ -stabilizers (e.g. niobium, molybdenum, vanadium, and tantalum) reduce it. Thus high  $\alpha$  alloys such as Ti-8Al-1V-Mo are prone to SCC, while IMI 679 is reputedly very resistant. For the relatively resistant Ti-6Al-4V alloy,  $\beta$  heat treatment lowers salt stress corrosion resistance. Although engines have been known to operate in salt-bearing atmospheres at stress and temperature levels in excess of those reported to cause cracking in laboratory tests, no hot salt corrosion problems with titanium alloys in service have been reported.<sup>54</sup> This may be due to dispersion of the corrosion products in service; also, the maintained concentration of the corrosive environment is likely to be much less than in laboratory tests.

Although it is generally accepted that the usefulness of titanium alloys is limited to about 550°C because of oxidation embrittlement, little attention has yet been paid to the development of more oxidation-resistant alloys (albeit with some reduction of high-temperature strength), or of protective coatings for known highly creep-resistant alloys. Efficient barriers to the diffusion of oxygen have yet to be discovered. Titanium alloys are very susceptible to galling, which can be reduced by wear-resistant coatings deposited, for example, by salt bath nitriding.<sup>33</sup> Fretting fatigue can be reduced by glass ball peening which introduces a compressively stressed skin.

Despite the well known pyrophoricity of titanium, the incidence of fires in engines has been low.<sup>54</sup> Titanium alloys can ignite and burn as a result of rubbing of components or entrapment of debris arising from mechanical failure elsewhere in the compressor. Design modifications and the use of abradable coatings on casings to clear rubbing surfaces have considerably reduced<sup>36</sup> the risk of titanium fires and ensured their containment within the engine.

Processes for the joining of titanium and its alloys are well established in the aero-engine industry. Owing to its marked reactivity with atmospheric gases, fusion welding of titanium is rigorously controlled,<sup>56, 57</sup> usually under a protective argon atmosphere. UK production experience includes not only pure and commercial titanium grades but also Ti-6Al-4V (IMI 318) and IMI 685. Electron beam welding is preferred for thin sections in order to minimize distortion. In the RB 211 engine, weight savings have been accomplished in the compressor by the use of drum construction, i.e. electron beam welded assemblies of prebalanced discs.<sup>58</sup> Resistance welding does not require a protective atmosphere. Diffusion bonding<sup>36</sup> has been used in the USA to fabricate hollow compressor blades in TMCA Ti-6Al-2Sn-4Zr-2Mo alloy.

The uncertainty and unpredictability of the aerospace industry is especially reflected in the fortunes of the titanium industry. So long as there is an energy crisis, whether real or threatened, the cost of production of such metals (i.e. in kWh/kg) must be a source of concern. Continued expansion of titanium utilization in engines depends both on enhancement of mechanical property characteristics and cost reductions. Although, in the USA, the price of mill products has been reduced from about \$36 to \$9 per kg in the period 1954-1972, component cost must be diminished by minimizing or eliminating machining. The use of castings is therefore economically attractive even though molten titanium, being highly reactive to conventional refractories, presents problems of mould containment. Precision casting processes have been developed. Disposable moulds<sup>59, 61</sup> are invariably used by impregnating a

conventional investment mould with carbon, although Russian processes have used centrifugal casting<sup>60</sup> and steel chill moulds. Tensile properties close to those of forgings in Ti-6Al-4V and Ti-5Al-2.5Sn alloys, but with somewhat reduced ductility and fatigue properties, have been obtained.<sup>57, 62</sup> There may be a need to machine away contaminated surface layers of the casting for fatigue-sensitive components, thereby reducing the economic advantage; the technology is still in an early stage. Preforms produced by pressing and sintering of Ti-6Al-4V powder are economically attractive, particularly for high surface area/weight ratio forgings.<sup>63</sup>

It is generally acknowledged that temperatures greater than about 550°C are excessive for titanium alloys, and certainly their most successful operation has been up to about 430°C. Nickel- and nickel-iron-based alloys are competitive in the range 430–540°C, even though they have lower specific fatigue properties and incur some weight penalty, particularly for discs. Their cost effectiveness depends not only on the component and the specific alloys considered but also on the country of manufacture. In the USA, titanium alloys tend to be somewhat cheaper than competitive nickel superalloys while the reverse has been true in the UK. Competition from steels and nickel-based alloys for the HP compressor may limit the role of titanium to operation below 400°C, i.e. where the weight-saving advantage of titanium alloys is more pronounced.<sup>64</sup>

The aero gas turbine industry, which is still the major beneficiary of the unique engineering properties of titanium, will be more concerned in the immediate future with the applicability of known alloys. As Hubbard<sup>39</sup> concludes in a recent review, there is a continuing need for a more thorough understanding of fatigue and toughness phenomena and optimization of process variables in order to achieve the highly desirable properties of titanium in large engineering components at minimum cost.

### Steels

In the early gas turbines, steels were widely used for turbine discs, the rims of which were uncooled. Good creep resistance required the use of complex austenitic steels such as G18B,<sup>65</sup> but even in the warm-worked condition the proof stress of this alloy was undesirably low, necessitating thick-hubbed designs. The disc weight was reduced by using heat-treatable martensitic steels, initially low-alloy types<sup>66</sup> such as Hecla 153 and HGT 3 (see Table VI), and subsequently steels of superior corrosion resistance based on S62 (13%Cr steel).<sup>65</sup> Their use, in association with rim cooling by compressor bleed air,

permitted disc temperatures up to about 400°C. With progressive increase in engine operating temperature, the need for improved creep resistance to about 450°C was satisfied by alloys containing carbide formers such as Mo, V, and Nb, e.g. H46,<sup>67</sup> Jethete M152<sup>68</sup> and FV 448<sup>69</sup> alloys. The latter was modified to give a further 50°C increase in creep performance by addition of boron and by solid solution strengthening with cobalt, resulting in FV 535 alloy (see Table VI and Figs.6 and 9). Hardwick<sup>70</sup> described the complexity of this alloy development, emphasizing that the real achievement was meeting the required properties consistently in the product form under production conditions. The maximum boron content was strictly controlled to avoid segregation problems in the air-melted ingots and to obtain low rejection rates during forging; the cobalt and chromium additions (see Table VI) were adjusted to avoid a tendency to embrittlement at temperatures about 100°C below the maximum service temperature. More consistent properties, as revealed by notched tensile tests, were obtained<sup>71</sup> by vacuum consumable arc remelting which necessitated a careful rebalancing of the C, Nb, and V contents.

The above range of martensitic 10–12%Cr steels has also been used in axial compressor discs and blades but, since the introduction of titanium alloys, steel has had a more limited role in such components, except in small helicopter engines where the excellent erosion resistance and formability of steels such as S62 are of paramount importance for compressor blading. However, stiffness (which often determines the weight of a dynamic component) and fatigue/ultimate strength ratio are attractively high to at least 600°C for compressor blade steels such as Jethete M152, FV 535, and FV 520B. FV 458 steel is used in compressor blades because of its good resistance to pitting corrosion. High stiffness, together with amenability to surface hardening by carburizing or nitriding, and good resistance to wear (i.e. to fretting and/or hammering) have made<sup>72</sup> certain steels uniquely useful for gears, bearings, and shafts. For compressor delivery casings, steels have for some time replaced light alloys. Fasteners<sup>72</sup> used in engines are largely made in steels of the 12%Cr martensitic type and in A286 alloy except where the greater creep strength of nickel-base alloys such as Nimonic 80A is needed.

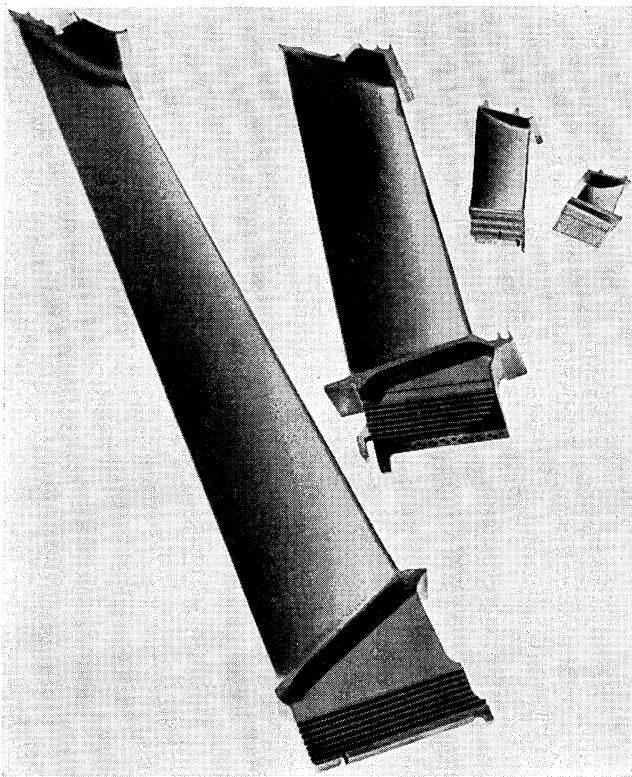
Although steels for discs and blades have been largely displaced by both nickel- and titanium-based alloys, the weight of steel in aero-engines is still considerable (see Table III). In the early versions of the RB 211 engine, the percentage of steels in the finished engine weight was close to 50%,<sup>72</sup> although on a volume basis it was about the same as that of titanium alloys,

TABLE VI  
Nominal Chemical Compositions and Densities of Steels

| Designation  | Composition, wt.-% |      |      |      |      |      |      |      |    |                       | Density,<br>10 <sup>3</sup> kg/m <sup>3</sup> |
|--------------|--------------------|------|------|------|------|------|------|------|----|-----------------------|-----------------------------------------------|
|              | C                  | Si   | Mn   | Cr   | Ni   | Mo   | Nb   | V    | Co | Others                |                                               |
| G18B         | 0.4                | —    | —    | 13   | 13   | 2.5  | 3.0  | —    | 10 | 2.5W                  | 8.13                                          |
| Hecla 153    | 0.4                | 0.3  | 0.5  | 1.1  | —    | 0.75 | —    | —    | —  | —                     | —                                             |
| HGT 3        | 0.25               | 0.4  | 0.4  | 3.0  | —    | 0.5  | —    | 0.75 | —  | 0.5W                  | 7.83                                          |
| S62          | 0.25               | 0.25 | 0.25 | 13.5 | 0.4  | —    | —    | —    | —  | —                     | 7.75                                          |
| H46          | 0.16               | 0.3  | 0.7  | 11.6 | —    | 0.6  | 0.25 | 0.3  | —  | —                     | 7.75                                          |
| Jethete M152 | 0.10               | 0.2  | 0.7  | 12   | 2.5  | 1.75 | —    | 0.3  | —  | 0.3N                  | 7.76                                          |
| FV 448       | 0.10               | 0.5  | 0.8  | 11   | 0.75 | 0.70 | 0.40 | 0.15 | —  | 0.05N                 | 7.76                                          |
| FV 458       | 0.06               | 2.6  | 3.2  | 12.5 | 4.25 | —    | 0.65 | —    | —  | —                     | 7.66                                          |
| FV 535       | 0.07               | 0.4  | 1.0  | 10.5 | 0.30 | 0.70 | 0.45 | 0.20 | 6  | —                     | 7.79                                          |
| FV 520B      | 0.05               | —    | —    | 14.3 | 5.5  | 1.6  | 0.35 | —    | —  | 1.8 Cu                | 7.83                                          |
| A286         | 0.06               | 0.65 | 1.6  | 15   | 25   | 1.25 | —    | 0.3  | —  | 2.1 Ti, 0.2Al, 0.003B | 7.93                                          |

i.e. about 20%. Steels are still potentially attractive on cost grounds for HP compressor and LP turbine discs, but, as Bennison and Gresham<sup>72</sup> point out, quality and inspection requirements are severe for engine components, particularly for large discs, and price is strongly governed by the rejection rate at late stages of manufacture. In industrial gas turbines where power/weight ratio considerations are relatively unimportant, the usage of steels in rotating components is governed more by cost and corrosion resistance requirements. Electroslag remelted forgings, e.g. for aero-engine shafts (3Cr-Mo-V steel) and industrial and marine gas turbine discs (FV 448), offer<sup>73</sup> freedom from the larger type of non-metallic inclusions and from conventional (air-melted) ingot segregation, and possess improved mechanical properties, particularly in the transverse direction. Lowe and Hogg<sup>74</sup> stated that, for aero gas turbine components, the extra cost of electroslag remelting is more than offset by the reduced rejection rate on final inspection.

It remains to be seen whether steels that are cost-attractive in finished component form, and retain the current advantages over nickel-base alloys of better hot and cold workability and greater ease of machining, can be developed with competitive creep strength much above 550°C and with acceptable fracture toughness, LCF strength, and corrosion resistance. The users' technical interest in steels is stimulated because of the extensive rig and engine test data on their crack propagation and crack tolerance characteristics,<sup>75</sup> and because of the good service experience of alloys such as FV 535. The challenge to the supplier to provide advanced steels that satisfy all the technical requirements in association with acceptable quality and cost is a considerable one.



10 View showing shape and size range of turbine blades

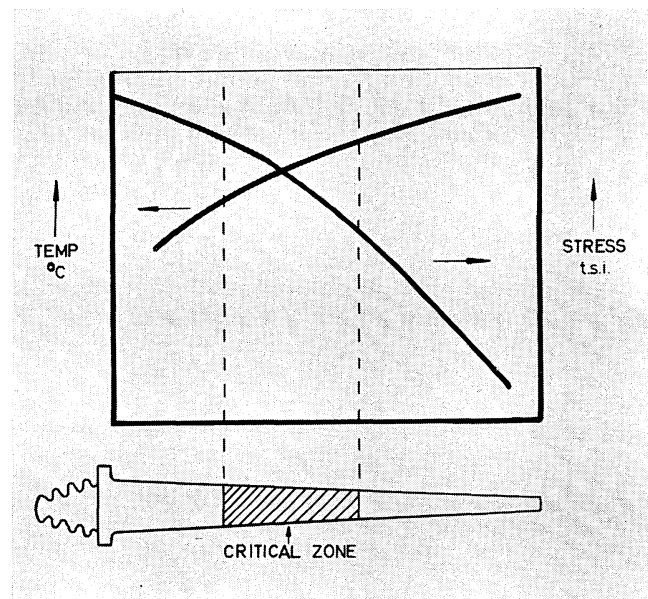
### Nickel- and Cobalt-Based Alloys

Except in the very early UK jet engines, nickel-base alloys in wrought, cast, or sheet form have been extensively used for turbine blading, combustion chambers, and nozzle guide vanes. In the USA, greater use was made of cobalt-base alloys in the aero gas turbines of the 1950's, stemming from turbo-supercharger experience, but this class of material is now largely confined to nozzle guide vanes, particularly in marine and industrial gas turbines. The superalloy content of engines (denoting nickel-, cobalt-, and nickel-iron-base creep-resistant alloys) has grown, particularly with the increase in the number of turbine stages in advanced engines and with the tendency to use nickel-base alloys for the final stages of HP compressor blades and discs as well as HP turbine discs. As shown in Table III, USA engines appear to have a greater content of superalloys than UK engines.

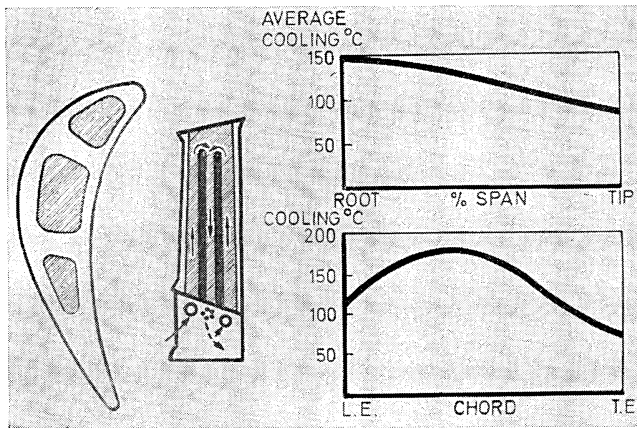
Because of the different performance requirements, alloys originally developed for turbine blades have often been modified for discs or for sheet metal constructions, such that the current trend is towards tailoring alloy development to specific applications. It is therefore convenient to review recent developments in superalloys in relation to the major engine components.

### Turbine Blade Alloys

Figure 10 shows four turbine blades ranging from about 3 to 28 cm in length according to the engine duty, i.e. from small shaft-power engines used in helicopters to the low-pressure turbine stage of the very large engines. The fir tree root design is common to most blades and it is important that the selected material satisfies the high-stress low-temperature requirements in this region as well as the lower-stress higher-temperature requirements of the integral aerofoil. Along the aerofoil there is a critical zone (see Fig.11) where the combination of stress and temperature is most severe. The stress range associated with this zone varies with the engine and blade design and blade stage, and, for a typical HP blade, varies between 120 and 200 MN/m<sup>2</sup> for a nickel-base alloy of



11 Stress-temperature distribution along blade aerofoil



12 Cooled HP turbine blade of Conway engine  
[Courtesy of Rolls-Royce (1971) Ltd]

density  $8.0 \times 10^3 \text{ kg/m}^3$ . Figure 12 indicates diagrammatically the simple cooling configuration of the HP turbine blade used in the RR Conway engine.

In UK production engines, nickel-base alloys have always been used for turbine blading. Their development for high-temperature high-strength applications was stimulated in the early 1940's by the need for constructional materials that would allow the aircraft gas turbine to achieve a satisfactory performance. The demands for creep resistance under the rather modest stresses and temperatures obtained in the prototype engines were satisfied by additions of 0.3%Ti and 0.1%C to an oxidation-resistant base, namely 80Ni–20Cr alloy, the resulting commercial alloy being Nimonic 75<sup>76</sup> (see Table VII). The need for higher engine speeds and for increase in the working temperature of engines provided the drive for intensive alloy development which has not yet abated. The work of Pfeil, Allen, and Conway<sup>77</sup> in the Mond Nickel Company (England) resulted in the development of Nimonic 80, a precipitation-hardened alloy in which the major strengthening phase consisted of a nickel–titanium–aluminium compound distributed in an oxidation-resistant nickel–chromium matrix. Investigation of compositional and heat treatment variables led to the development of the Nimonic series<sup>76, 78</sup> with progressive increases in creep strength, achieved by a combination of solid solution strengthening and precipitation hardening. The substitution of 20%Co for the nickel led to Nimonic 90, the cobalt serving to raise the solubility of the precipitation-hardening phase in the matrix and to improve workability. Increases in the level of aluminium and titanium contents were made possible by vacuum melting techniques and improved methods of hot working, and these advances gave rise to Nimonic 95. Molybdenum additions with some reduction of chromium content resulted in Nimonic 105. Although the development of such alloys was largely empirical, the relationship between the chemical constitution, micro-structure, and mechanical properties of these alloys has since been intensively studied and is generally if not completely understood. By about 1960 a series of hot-worked alloys had not only been developed but successfully used in British aero-engines.

After 1960, the introduction of air cooling was a direct and effective way of increasing the cycle temperature (see Fig.4) without the need for major advances in alloy development. In the USA, the benefits of cooling technology were not appreciated as early as in the UK, and high-temperature alloy devel-

opment was therefore more intensively pursued. Hot working of the blade material became more difficult with increase in its high-temperature strength such that, in the case of Nimonic 115 and Nimonic 120 with a titanium and aluminium content of 8–9%, extrusion rather than forging was necessary to break down the cast structure. Further alloying additions were effective only if the alloys were used in the cast state, and the blade was cast virtually to finished shape. Engine manufacturers on both sides of the Atlantic now appreciate the merits of combining advanced cooling technology and advanced cast alloys.

During the past ten years the number and complexity of available blade alloys have increased enormously, to the point where alloys containing twelve constituent elements, each having a significant role, are not uncommon. Nickel, often with cobalt and sometimes iron in solid solution, forms the austenitic face-centred cubic matrix commonly termed the 'gamma' phase (the cobalt serves to control the structure and restrict the precipitation of unwanted phases while iron appears only in the less expensive alloys). The electropositive elements aluminium and titanium form with the highly electronegative nickel a coherent intermetallic precipitate,  $\text{Ni}_3(\text{Ti}, \text{Al})$ , commonly called 'gamma prime', which provides the major strengthening phase. Chromium gives surface stability by reacting with elements such as oxygen and sulphur in the engine combustion gases, thereby helping to protect the alloy from oxidation and hot corrosion; it also strengthens the alloy by forming carbides. The small reactive atoms, carbon and boron, form hard carbides and borides that serve to control the grain boundary structure, probably by filling discontinuities at grain boundaries and so strengthening the bond, but their precise role is still obscure. The refractory metals niobium, molybdenum, tantalum, and tungsten strengthen both the  $\gamma$  solid solution phase and the  $\gamma'$  phase and also appear as carbides. They serve to increase the temperature at which the  $\gamma'$  phase goes into solid solution and thereby permit the alloy to operate at higher temperatures. The refractory metal content can be as high as 15% by weight in cast alloys. Greater amounts can be self-defeating because additions of high-density elements such as tungsten and tantalum increase the density of the alloy, and so its specific strength, which is the most important parameter in centrifugally loaded components, may be unaffected or even diminished.

In the USA the philosophy<sup>79, 80</sup> of wrought nickel-based alloy development has been similar to that in the UK, the alloys Udimet 500 and 700<sup>81</sup> (see Table VII) being equivalent in creep-rupture properties to Nimonics 105 and 115 (see Fig.13) and of broadly similar composition. Inconel 718,<sup>82</sup> an air-melted niobium-containing alloy, the  $\gamma'$  phase being  $\text{Ni}_3\text{Nb}$ , is widely used for blade, disc, and sheet applications. Unitemp AF2-1D<sup>83</sup> is claimed to have better creep strength than any other commercial hot-worked alloy. As in the UK, further improvements in creep strength are unlikely because of limitations on hot workability. Because of the greater use of marine and industrial gas turbines which operate in more corrosive environments for longer times than aero-engines, the trend in wrought alloy development has been to sacrifice some strength for improved metallurgical stability and corrosion resistance (see below). Although this trend applies also to cast alloys, the intensive development of the latter in the USA and to a lesser extent in the UK has been motivated by aero-engine needs and aided by the growth and sophistication of vacuum melting technology which has given greater compositional control of impurities and of readily oxidizable elements such as titanium. Cast alloys such as Inconel 713C

and GMR 235 showed a temperature advantage of about 50°C over contemporary hot-worked alloys; subsequently a low-carbon version (713LC) of the former alloy was developed<sup>84</sup> with increased ductility, permitting better stress redistribution in cast-integrally-bladed turbine discs. The British alloy,

G64,<sup>65</sup> was successfully used in Proteus engines powering Britannia aircraft. An even stronger alloy, IN 100, was developed<sup>85</sup> in the USA in 1960 and is widely used in military engines. It was selected<sup>86</sup> for the cooled HP rotor blading of the Olympus 593 and subsequently the Pegasus engines. Its

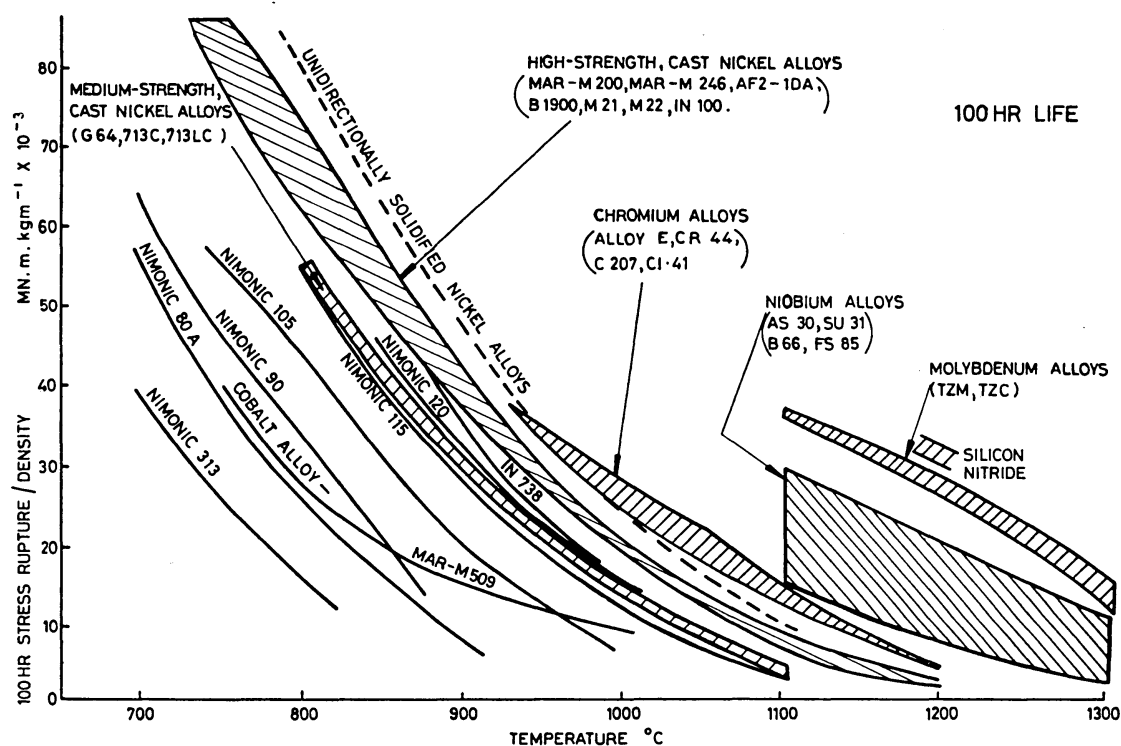
TABLE VII  
Nominal Chemical Compositions and Densities of Nickel-Base Alloys for Turbine Blades and Nozzle Guide Vanes

| Alloy             | Composition, wt-%<br>C Cr Co Al Ti Mo W Nb Ta |      |           |     |     |      |      |     |         |   | Fe       | B          | Zr        | Others                                                               | Density,<br>10 <sup>3</sup> kg/m <sup>3</sup> |
|-------------------|-----------------------------------------------|------|-----------|-----|-----|------|------|-----|---------|---|----------|------------|-----------|----------------------------------------------------------------------|-----------------------------------------------|
| Nimonic 75        | 0.10                                          | 19.5 | —         | —   | 0.4 | —    | —    | —   | —       | — | 5.0      | —          | —         | —                                                                    | 8.37                                          |
| Nimonic 80A       | 0.10                                          | 19.5 | 2.0       | 1.4 | 2.5 | —    | —    | —   | —       | — | max. 3.0 | 0.008      | 0.015     | —                                                                    | 8.19                                          |
| Nimonic 90        | 0.13                                          | 19.5 | max. 18.0 | 1.5 | 2.4 | —    | —    | —   | —       | — | max. 1.5 | max. 0.02  | max. 0.15 | —                                                                    | 8.18                                          |
| Nimonic 95        | 0.15                                          | 19.5 | 18.0      | 2.0 | 2.9 | —    | —    | —   | —       | — | max. 5.0 | —          | —         | —                                                                    | 8.11                                          |
| Nimonic 100       | 0.30                                          | 11.0 | 20.0      | 5.0 | 1.5 | 5.0  | —    | —   | —       | — | 2.0      | —          | —         | —                                                                    | 8.04                                          |
| Nimonic 105       | 0.12                                          | 14.8 | 20.0      | 4.7 | 1.2 | 5.0  | —    | —   | —       | — | max. 1.0 | 0.007      | 0.15      | —                                                                    | 8.01                                          |
| Nimonic 108       | 0.20                                          | 14.8 | 20.0      | 4.7 | 1.2 | 5.0  | —    | —   | —       | — | max. 2.0 | 0.007      | 0.15      | —                                                                    | 8.01                                          |
| Nimonic 115       | 0.15                                          | 15.0 | 14.8      | 5.0 | 4.0 | 4.0  | —    | —   | —       | — | max. 1.0 | 0.018      | 0.15      | —                                                                    | 7.85                                          |
| Nimonic 120       | 0.04                                          | 12.5 | 10.0      | 4.5 | 3.5 | 5.7  | —    | —   | —       | — | max. 0.4 | 0.03       | 0.05      | —                                                                    | 7.97                                          |
| M21 (PD 21)       | 0.10                                          | 5.8  | 1.0       | 6.0 | 0.5 | 2.0  | 10.5 | 1.5 | —       | — | max. 0.5 | max. 0.02  | max. 0.12 | —                                                                    | 8.53                                          |
| M22               | 0.13                                          | 5.7  | —         | 6.3 | —   | 2.0  | 11.0 | —   | 3.0     | — | max. 1.0 | —          | 0.60      | —                                                                    | 8.63                                          |
| M313 (Nimonic 81) | 0.06                                          | 30.0 | —         | 0.9 | 1.7 | —    | —    | —   | —       | — | max. 2.0 | 0.003      | 0.05      | —                                                                    | 8.06                                          |
| MAR-M200          | 0.15                                          | 9.0  | 10.0      | 5.0 | 2.0 | —    | 12.5 | 1.0 | —       | — | —        | 0.015      | 0.05      | —                                                                    | 8.52                                          |
| MAR-M246          | 0.15                                          | 9.0  | 10.0      | 5.5 | 1.5 | 2.5  | 10.0 | —   | 1.5     | — | —        | 0.015      | 0.05      | —                                                                    | 8.44                                          |
| MAR-M002          | 0.15                                          | 9.0  | 10.0      | 5.5 | 1.5 | 10.0 | —    | —   | 2.5     | — | 1.0      | 0.015      | 0.05      | 1.5Hf                                                                | —                                             |
| MAR-M004          | 0.05                                          | 12.0 | —         | 5.9 | 0.6 | 4.5  | —    | 2.0 | —       | — | —        | 0.010      | 0.10      | 1.3Hf                                                                | —                                             |
| MAR-M007          | 0.10                                          | 8.0  | 10.0      | 6.0 | 1.0 | 6.0  | —    | —   | 2.5     | — | —        | 0.015      | 0.08      | 1.3Hf                                                                | —                                             |
| B1900             | 0.10                                          | 8.0  | 10.0      | 6.0 | 1.0 | 6.0  | —    | —   | 4.3     | — | 0.35     | 0.015      | 0.08      | —                                                                    | 8.22                                          |
| G64               | 0.13                                          | 11.0 | —         | 6.0 | —   | 3.0  | 4.0  | 2.0 | —       | — | —        | 0.25       | —         | —                                                                    | 8.00                                          |
| GMR 235           | 0.15                                          | 15.5 | —         | 3.0 | 2.0 | 5.3  | —    | —   | —       | — | 10.0     | 0.05       | —         | —                                                                    | 8.03                                          |
| C242              | 0.30                                          | 20.5 | 10.0      | 0.2 | 0.3 | 10.5 | —    | —   | —       | — | 1.0      | —          | —         | 0.3Mn, 0.3Si                                                         | 8.40                                          |
| C1023             | 0.12                                          | 14.5 | 9.0       | 3.9 | 3.4 | 7.5  | —    | —   | —       | — | —        | 0.004      | —         | —                                                                    | —                                             |
| HRA 376           | 0.05                                          | 6.5  | 11.5      | 5.9 | —   | —    | 8.5  | —   | 8.5     | — | max. 0.1 | 0.02       | 0.10      | —                                                                    | 8.85                                          |
| IN 591-X          | 0.10                                          | 3.0  | 12.0      | 5.7 | —   | —    | 19.0 | —   | 3.0     | — | —        | 0.03       | 0.37      | —                                                                    | 9.13                                          |
| Udimet 500        | 0.15                                          | 17.5 | 16.5      | 2.9 | 2.9 | 4.0  | —    | —   | —       | — | 4.0      | 0.010      | —         | —                                                                    | 8.02                                          |
| Udimet 700        | 0.15                                          | 15.0 | 17.0      | 4.3 | 3.4 | 5.3  | —    | —   | —       | — | max. 4.0 | max. 0.05  | —         | —                                                                    | 7.91                                          |
| Unitemp AF2-1DA   | 0.35                                          | 12.0 | 10.0      | 4.5 | 3.0 | 3.0  | 6.0  | —   | 1.5     | — | max. 0.5 | max. 0.015 | 0.1       | —                                                                    | 8.08                                          |
| WAZ-20            | 0.15                                          | —    | —         | 6.2 | —   | —    | 18.5 | —   | —       | — | —        | —          | 1.5       | —                                                                    | 9.02                                          |
| TAZ-8B            | 0.13                                          | 6.0  | 5.0       | 6.0 | —   | 4.0  | 4.0  | 1.5 | 8.0     | — | —        | 0.004      | 1.0       | —                                                                    | —                                             |
| IN 713LC          | 0.12                                          | 12.5 | —         | 6.1 | 0.8 | 4.2  | —    | 2.2 | (Nb+Ta) | — | —        | 0.012      | 0.10      | —                                                                    | 7.91                                          |
| IN 713C           | 0.05                                          | 12.0 | —         | 5.9 | 0.6 | 4.5  | —    | 2.0 | (Nb+Ta) | — | 0.5      | 0.01       | 0.10      | —                                                                    | 8.01                                          |
| IN 100            | 0.18                                          | 10.0 | 15.0      | 5.5 | 4.7 | 3.0  | —    | —   | —       | — | —        | 0.014      | 0.06      | 1.0V                                                                 | 7.75                                          |
| IN 587            | 0.05                                          | 28.5 | 20.0      | 1.2 | 2.3 | —    | —    | 0.7 | —       | — | —        | 0.003      | 0.05      | —                                                                    | 8.08                                          |
| IN 738            | 0.17                                          | 16.0 | 8.5       | 3.4 | 3.4 | 1.8  | 2.6  | 0.9 | 1.8     | — | 0.5      | 0.010      | 0.10      | —                                                                    | 8.11                                          |
| IN 792            | 0.21                                          | 12.7 | 9.0       | 3.2 | 4.2 | 2.0  | 3.9  | —   | 3.9     | — | max. 0.2 | 0.02       | 0.10      | Optional 0.8Hf                                                       | 8.25                                          |
| IN 853            | 0.05                                          | 20.0 | —         | 1.5 | 2.5 | —    | —    | —   | —       | — | —        | 0.007      | 0.07      | 1.3Y <sub>2</sub> O <sub>3</sub> , 1.2Al <sub>2</sub> O <sub>3</sub> | 8.09                                          |
| IN 597            | 0.5                                           | 24.5 | 20.0      | 1.5 | 3.0 | 1.5  | —    | 1.0 | —       | — | —        | 0.012      | 0.05      | —                                                                    | 8.04                                          |
| Cosint 1000       | 0.3                                           | 16.0 | 20.0      | 5.5 | 2.5 | 2.5  | 5.5  | —   | —       | — | —        | 0.06       | 0.2       | —                                                                    | —                                             |
| NX 188            | 0.04                                          | —    | —         | 8.0 | —   | 18.0 | —    | —   | —       | — | —        | —          | —         | —                                                                    | 8.19                                          |
| EI 437B           | 0.08                                          | 20.0 | —         | 0.7 | 2.6 | —    | —    | —   | —       | — | —        | 0.005      | —         | —                                                                    | 8.19                                          |
| EI 867            | 0.1                                           | 9.0  | 5.0       | 4.5 | —   | 10.3 | 5.0  | —   | —       | — | 5.0      | 0.02       | —         | 0.6V, 0.02C <sub>max</sub> , 0.1Ba <sub>max</sub>                    | —                                             |
| EI 929            | max. 0.12                                     | 10.5 | 14.0      | 4.0 | 1.7 | 5.0  | 5.5  | —   | —       | — | max. 5.0 | max. 0.02  | —         | 0.6V, 0.1Ba <sub>max</sub>                                           | —                                             |
| ZHS 6-K           | 0.16                                          | 11.0 | 4.5       | 5.5 | 2.8 | 4.0  | 5.0  | —   | —       | — | max. 2.0 | 0.02       | —         | —                                                                    | —                                             |

compositional limits have since been modified and more closely controlled to render it structurally stable at elevated temperatures after the long durations associated with civil as opposed to military use. The alloys M21 and M22, developed in the UK by International Nickel Ltd, were designed to give high rupture strength by virtue of their low chromium and high refractory metal content.<sup>87, 88</sup> These alloys were necessarily of relatively high density and, on a specific strength basis, were not greatly superior to IN 100 (see Fig.13). At about the same time, Martin Metals Corporation introduced a high-tungsten, low-chromium alloy, MAR-M200,<sup>89</sup> of similar rupture strength to M21, and subsequently MAR-M246,<sup>90</sup> in which the tungsten was partly replaced by molybdenum and tantalum in order to provide better long-term stability and improved properties in the 'ductility trough', i.e. at 700°–800°C where the tensile ductility of precipitation-hardened superalloys is lowest. The latter alloy has been recently modified by the addition of 1.5% hafnium (i.e. M002 alloy) which is claimed to improve strength and ductility and reduce the sensitivity of mechanical properties to changes in section thickness.<sup>91</sup> Hafnium additions to alloys such as B1900 (favoured<sup>92</sup> for recent Pratt-Whitney engines), 713LC, and Udimet 700 have refined and varied the composition of the MC carbides, significantly increased the volume of  $\gamma/\gamma'$  eutectic and altered the morphology of the  $\gamma'$  phase.<sup>91</sup> It has also appreciably increased the cost of these alloys. NASA's development programme has emphasized the use of refractory metal additions, notably tantalum. Such alloys, e.g. TAZ 8B,<sup>93</sup> HRA 376,<sup>86</sup> and IN 591X,<sup>87</sup> are relatively heavy and expensive and some may have limited hot corrosion resistance; their high melting points may be more advantageous for nozzle guide vanes where the weight penalty is less serious than for rotor blades. Alloy IN 792,<sup>94</sup> containing tantalum and,

optionally, hafnium, has excellent hot strength and corrosion resistance.

After World War II the requirement in the USA for precision-cast turbine blades stimulated<sup>94</sup> the development of the cobalt-based alloy HS 21 (see Table VIII), a modification of Vitallium with lower carbon content, and an addition of nickel to improve ductility. The high chromium content, typically 25%, is needed for oxidation resistance. Increase in the nickel content and replacing the molybdenum in this alloy by tungsten led to the alloy X40 (HS 31). Both these air-cast alloys can withstand moderate loads for long periods at high temperature, and their inherent stability coupled with good corrosion and thermal shock resistance made them more suitable for the higher-temperature and lower-stress conditions encountered in nozzle guide vanes and afterburners. The newer, more highly alloyed MAR-M alloys, which contain titanium, zirconium, and tantalum, as well as tungsten for strengthening, are vacuum melted<sup>95</sup> (vacuum melting and casting of alloys such as HS 31 and WI 52 had little effect on their mechanical properties). The best of the recently developed alloys is probably MAR-M509 which combines the good oxidation resistance of MAR-M302 with the attractive rupture strength of MAR-M322. Cobalt-based superalloys rely on two main strengthening mechanisms<sup>188</sup>: solid solution strengthening by additions of the refractory metals, and precipitation hardening by carbides of these metals and chromium. An effective stable precipitate of an intermetallic compound coupled with carbide strengthening has still to be developed<sup>96</sup> for cobalt-base alloys. Although the melting point of cobalt is some 70°C higher than that of nickel, cobalt-base alloys are generally weaker than nickel-base alloys at temperatures of 750°–900°C and are therefore rarely used for first-stage turbine blading; however, alloys such as S-816 and HS 31 have been used for



13 Specific 100 h rupture stress v. temperature for turbine blade materials

TABLE VIII  
Nominal Chemical Compositions and Densities of Cobalt-Base Alloys

| Alloy        | Composition, wt-% |      |          | Mo  | W    | Ta        | Ti   | Al  | B          | Zr   | Fe        | Si       | Mn        | Others | Density, 10 <sup>3</sup> kg/m <sup>3</sup> |
|--------------|-------------------|------|----------|-----|------|-----------|------|-----|------------|------|-----------|----------|-----------|--------|--------------------------------------------|
| Vitallium    | 0.2               | 30   | —        | 6   | —    | —         | —    | —   | —          | —    | —         | —        | —         | —      | 8.10                                       |
| FSX 414      | 0.25              | 29.5 | 10.5     | —   | 7    | —         | —    | —   | 0.012      | —    | 2 max.    | 1 max.   | 1 max.    | —      | 8.30                                       |
| FSX 418      | 0.25              | 29.5 | 10.5     | —   | 7    | —         | —    | —   | 0.012      | —    | 2 max.    | 1 max.   | 1 max.    | 0.15Y  | —                                          |
| HS 21        | 0.25              | 27   | 2.75     | 5.5 | —    | —         | —    | —   | 0.007 max. | —    | 3 max.    | —        | —         | —      | 8.30                                       |
| HS 31 (X 40) | 0.5               | 25.5 | 10.5     | —   | 7.5  | —         | —    | —   | —          | —    | 2 max.    | 1 max.   | 1 max.    | —      | 8.61                                       |
| WI 52        | 0.45              | 21   | 1.0 max. | —   | 11   | 2 (Ta+Nb) | —    | —   | —          | —    | 1.75 max. | 0.5 max. | 0.5 max.  | —      | 8.88                                       |
| MAR-M 302    | 0.85              | 21.5 | —        | —   | 10   | 9         | —    | —   | 0.005      | 0.2  | 1.5 max.  | 0.4 max. | 0.2 max.  | —      | 9.22                                       |
| MAR-M 322    | 1.0               | 21.5 | —        | —   | 9    | 4.5       | 0.75 | —   | —          | 2.25 | 0.7       | 0.2      | 0.1       | —      | 8.91                                       |
| MAR-M 509    | 0.6               | 23.5 | 10       | —   | 7    | 3.5       | 0.2  | —   | —          | 0.5  | 1.5 max.  | 0.4 max. | 0.1 max.  | —      | 8.86                                       |
| S-816        | 0.38              | 20   | 20       | 4   | 4    | 4Nb       | —    | —   | —          | —    | 4         | —        | 1.2       | —      | 8.66                                       |
| AiResist 213 | 0.18              | 19   | —        | —   | 4.7  | 6.5       | —    | 3.5 | —          | 0.15 | —         | —        | —         | 0.1Y   | 8.50                                       |
| Haynes 188   | 0.10              | 22   | 22       | —   | 14.5 | —         | —    | —   | —          | —    | 3 max.    | 0.35     | 1.25 max. | 0.09La | 9.13                                       |

the second and third turbine stages of USA but not UK aero-engines. The recent stronger cobalt-base alloys have lower chromium contents than X40. With the development of cast nickel-based alloys such as IN 738,<sup>97</sup> of comparable strength and hot corrosion resistance (*see below*) and potentially lower cost, it is difficult to envisage any future role for such precipitation-hardened cobalt-base alloys in rotating parts, even of non-aero gas turbines.

Since the introduction of casting alloys, increasing attention has been paid to control of the casting process. Conventionally, blades in air-cast or vacuum-cast nickel-base alloys are made by shell moulding techniques such that the grain structure is random or equiaxed. In order to obtain high-quality castings, attention has been paid not only to control of the alloying additions and trace impurities (*i.e.* cleanliness of the melting stock), but also to control of the reactivity of the molten alloy with crucible and mould refractories. The creep-rupture properties are markedly influenced<sup>98</sup> by grain size and grain orientation. Improved thermal fatigue resistance has been obtained<sup>99</sup> with castings containing fine surface grains which can be controlled by adding suitable nucleants<sup>100</sup> to the primary mould coating or to the actual melt.<sup>76</sup> Stress-rupture properties vary with section size and shape, low lives being associated with thin sections and sometimes, but not invariably, with the number of grains in the section. Hockin and Taylor<sup>101</sup> computerized the heat transfer characteristics of a blade casting and the surrounding mould in order to obtain thermal histories for various parts of the casting, and they established that the strength levels at any point in the casting

could be related to those of test bars with similar thermal histories.

The importance of grain orientation is reflected in the service failure of conventionally cast blades, which invariably occurs by creep-rupture in grain boundaries transverse to the direction of centrifugal loading. Ver Snyder appreciated<sup>102</sup> that improved blade life could result from elimination of such boundaries by directional solidification, *i.e.* by creating a steep temperature gradient in the mould during casting, such that the solid/liquid interface at the growing grains moved in the direction of heat flow. The required gradient can be created by pouring the molten metal into a mould heated above the metal temperature and mounted on a water-cooled copper chill. As freezing occurs from the chill, columnar grain growth is induced in the vertical direction.

Pioneering work by Ver Snyder, Pearcey, and co-workers showed<sup>103–105</sup> that a directionally solidified nickel-based alloy (MAR-M200) had superior strength and ductility particularly at about 700°C to the conventionally cast alloy, and also a lower modulus of elasticity. The thermal fatigue resistance of turbine blades and vanes was thereby markedly increased, and in engine tests increased blade life and reduced distortion of vanes were evident.<sup>106, 107</sup> It is claimed<sup>108</sup> that directionally solidified alloys are used by Pratt and Whitney in the TF 30 and F100 military engines and that FAA approval has been obtained for their use in turbine blades in the JT9D-7A engine in order to permit a higher TET and achieve the highest thrust for powering the high gross weight versions of the Boeing 747 transport. Northwood and Homewood,<sup>109</sup> using a moving

TABLE IX  
Effect of Directional Solidification on Stress-Rupture Lives of MAR-M200 Type Alloy

| Designation | Condition                                 | Creep-rupture properties     |         |                         | 982°C, 206 MN/m <sup>2</sup> |               |                         |
|-------------|-------------------------------------------|------------------------------|---------|-------------------------|------------------------------|---------------|-------------------------|
|             |                                           | 760°C, 690 MN/m <sup>2</sup> | Life, h | Elongation, %           | Life, h                      | Elongation, % | Min. creep rate per h   |
| PWA 659     | Conventionally cast, equiaxed             | 4.9                          | 0.45    | 70.0 × 10 <sup>-5</sup> | 35.6                         | 2.6           | 23.8 × 10 <sup>-5</sup> |
| PWA 664     | Directionally solidified, polycrystalline | 366.0                        | 12.6    | 14.5 × 10 <sup>-5</sup> | 67.0                         | 23.6          | 25.6 × 10 <sup>-5</sup> |
| PWA 1409    | Directionally solidified, single-crystal  | 19140                        | 14.5    | 2.2 × 10 <sup>-5</sup>  | 107.0                        | 23.6          | 16.1 × 10 <sup>-5</sup> |

furnace technique, showed that the properties of air-cast 713C alloy could be improved by directional solidification, especially at higher rates of solidification than those used in previous USA work. No further enhancement of stress-rupture properties was obtained by vacuum processing.<sup>110</sup> Subsequent USA work, using a fixed furnace-moving mould technique, confirmed the benefits of high solidification rates (i.e. 25–38 cm/h).

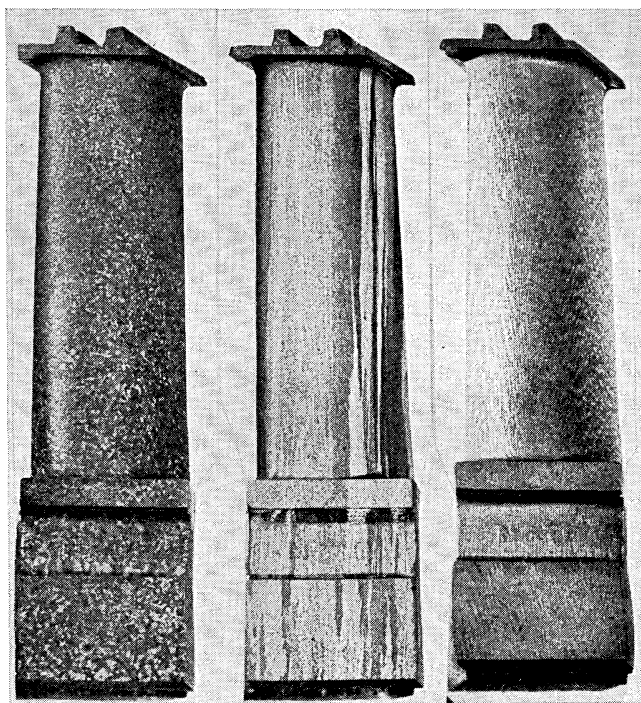
The directional solidification process resulted in preferred growth of the  $\langle 001 \rangle$  orientation and, by suitable mould design, e.g. by placing suitable constrictions in the path of the columnar grains, one of them could grow selectively to produce a single-crystal blade. Table IX shows the respective merits of random (PWA 659), polycrystalline columnar (PWA 664), and single-crystal (PWA 1409) structures obtained in one alloy composition by the United Aircraft Laboratories. Figure 14 shows the three representative structures obtained in turbine blades vacuum-cast in MAR-M246 alloy at the National Gas Turbine Establishment, UK. Recent work has been concerned with establishing production procedures that are economically viable and with determining the extent to which the process improves the properties, particularly the creep-rupture strength (see Fig.15) of more recently developed alloys. The latter studies have included investigating the effects of heat treatment, the properties of thin sections, and the mechanisms of strengthening.<sup>111</sup> Liquid metals such as tin have been used<sup>137</sup> to provide the desired temperature gradient. Little effort has been devoted to optimizing compositions which would benefit more from controlled solidification than alloys developed for conventional casting processes and which would possess the most desirable combination of strength and ductility.

The USSR adopted Nimonic 80A (i.e. EI437) but subsequent alloy development proceeded independently. Shortage

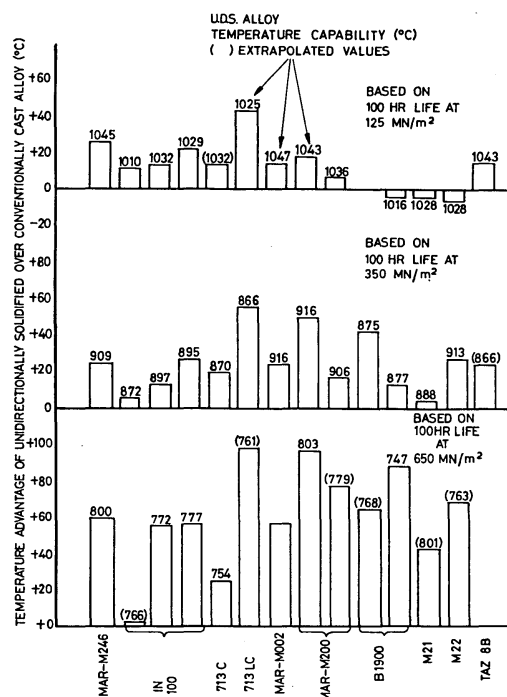
of molybdenum and cobalt led to the use of higher tungsten content than was used in the alloys developed in the West during the 1950's.<sup>112</sup> Cobalt was not introduced as an alloying addition until about 1959. Although their aluminium plus titanium contents are lower, the available data<sup>48, 113, 114</sup> on Russian alloys such as the cobalt-containing EI 867 and EI 929 suggest that their high-temperature strength is comparable with UK and USA hot-worked alloys.

Russian-cast alloys, e.g. ZHS 6-K, comparable in rupture strength to IN 100 and MAR-M200 are known. From exhibits at recent Paris Air Shows it is clear that the USSR is able to produce cooled blades and directionally solidified structures.<sup>115</sup> The French aero-engine industry has relied heavily on USA and UK blading alloys, but is now taking a more active interest in alloy and associated process development. Other European countries such as Germany, Italy, and Sweden have preferred to produce USA and UK aero-engines under licence and therefore have had little incentive to pursue alloy development programmes.

In comparison with hot working and casting processes, only modest attempts have been made to utilize powder metallurgy techniques for making precipitation-hardened turbine blade alloys. In the UK, cold mechanical and isostatic pressing of air-atomized powders followed by vacuum sintering has been employed. Sintered cobalt-based alloys (Vitallium)<sup>116</sup> and nickel-based alloys (e.g. Cosint 1000<sup>117</sup>) and alloys of similar<sup>118</sup> composition to Nimonic 100 and 115, possessing up to 98–99% theoretical density and a more uniform grain size than wrought alloys have been prepared and evaluated. Limited engine testing of blades machined from sintered blocks or oversize aerofoil blocks was encouraging.<sup>118</sup> However, as an economically competitive route for producing alloys of similar strength to hot-worked alloys, powder metallurgy techniques have not been thoroughly investigated, partly because of technical



14 Conventionally cast (left), and directionally solidified, polycrystalline (centre) and single-crystal (right) blades cast at NGTE



15 Temperature advantage obtained by directional solidification of various superalloys

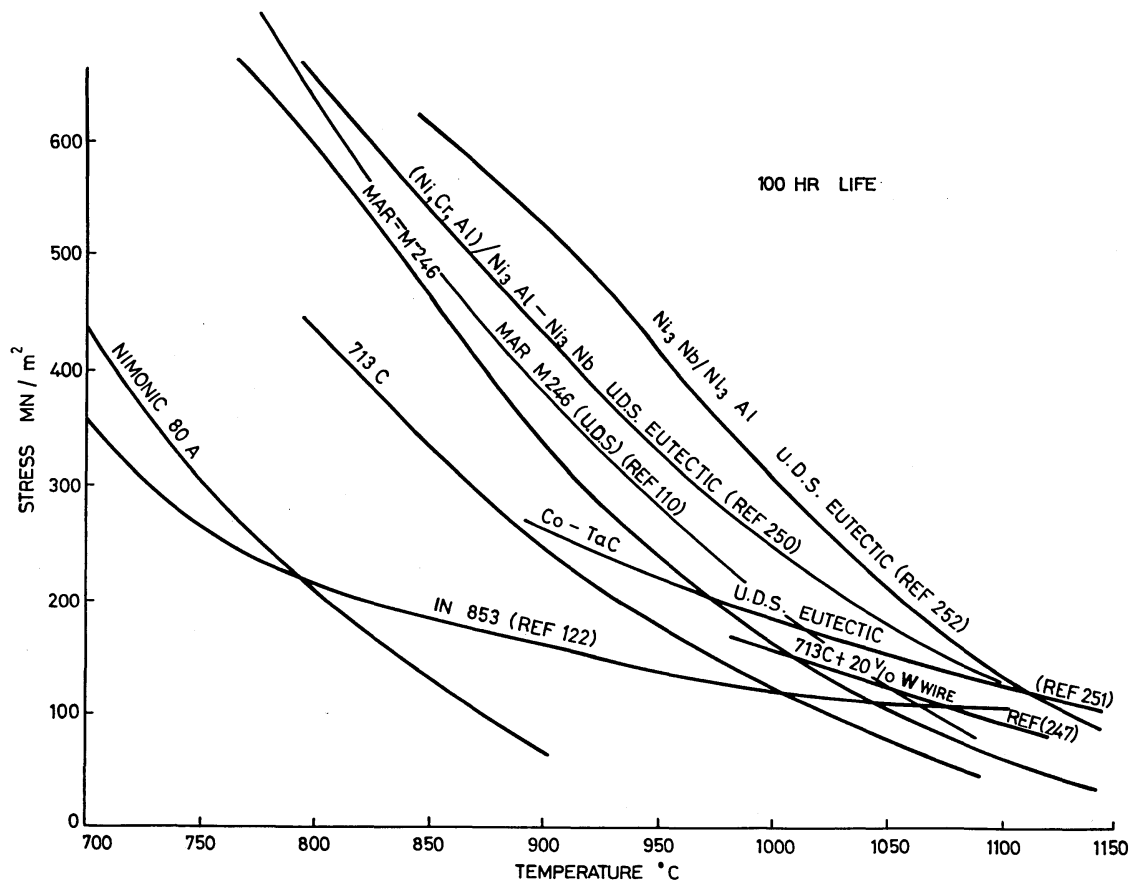
deficiencies in the critical properties. For example, sintered alloys of similar composition and rupture strength to Nimonic 115 have been developed with adequate fatigue strength, but significantly lower tensile ductility and resistance to impact and to thermal fatigue.<sup>118</sup>

Two approaches that would overcome the limitations of precipitation hardening have been examined. The first approach of dispersion strengthening using powder metallurgy techniques and mechanical mixing was examined<sup>119</sup> some twenty years ago. In 1962 much finer Ni-2ThO<sub>2</sub> powders were produced by a chemical codeposition technique; the uniform submicron dispersion of thorium after controlled thermomechanical working of the sintered product was effective<sup>120</sup> in retaining strength to higher temperatures. Relative to precipitation-hardened alloys, TD Ni had the advantages of a higher melting point (similar to pure nickel), higher thermal conductivity, and freedom from intergranular oxidation. However, its level of general oxidation resistance was little better than that of pure nickel and so TD NiCr<sup>121</sup> (80Ni-20Cr-2ThO<sub>2</sub>) was developed, giving superior oxidation resistance to TD Nickel above 900°C but possessing somewhat inferior rupture strength (see Fig.16). Unfortunately the strength level of both these alloys was not adequate at low and intermediate temperatures for turbine blades. All attempts to introduce the thorium by chemical or conventional mechanical techniques into a creep-resistant alloy failed to provide the desired uniform distribution and interparticle spacing of less than half a micron. Recently it was demonstrated that a high-speed (or

high-energy) mechanical blending technique could be effective<sup>122, 123</sup> in intimately mixing yttria and pre-alloyed Nimonic 80A powders; alternate cold welding and fragmentation took place until all the constituents were finally divided and uniformly distributed. The yttriated alloy possessed similar rupture strength to Nimonic 80A at intermediate temperatures yet retained its strength much better at higher temperatures (see Fig.16). It remains to be seen whether superalloys of present-day strength levels can be dispersion strengthened to the same degree by this technique. The second approach, fibre strengthening, is described below (see 'Fibre Composites').

#### Nickel-Base Alloys for Discs

With increase in TET and the consequent increase in disc rim and hub temperatures, steels such as A286 and FV 535 have been superseded by nickel-iron- and nickel-based alloys for the high-pressure compressor and turbine disc stages. In the USA, Ni-Fe-Cr alloys (see Table X and Figs.8 and 9) such as Incoloy 901<sup>124</sup> (precipitation hardened by alloying with titanium and aluminium, and solid solution strengthened with molybdenum) and Inconel 718 (strengthened by the Ni<sub>3</sub>(Ti, Al, Nb)  $\gamma'$ -phase) have found widespread use despite increased cost and reduced forgeability and machineability relative to steels. Careful control of melting practice, involving vacuum melting followed by either vacuum remelting or electroslag refining, has ensured greater ease of forgeability and better uniformity of properties, particularly for the larger discs. The importance of specifying in detail the melting and



16 Comparative rupture strength merits of precipitation-, dispersion-, and fibre-strengthened alloys

TABLE X  
Nominal Chemical Compositions and Densities of Nickel-Base Alloys for Turbine and Compressor Discs

| Alloy         | Composition, wt-% |      |      | Al   | Ti  | Mo   | W   | Nb  | Fe       | B     | Zr    | Density, 10 <sup>3</sup> kg/m <sup>3</sup> |
|---------------|-------------------|------|------|------|-----|------|-----|-----|----------|-------|-------|--------------------------------------------|
|               | C                 | Cr   | Co   |      |     |      |     |     |          |       |       |                                            |
| Incoloy 901   | 0.05              | 12.5 | —    | 0.25 | 2.8 | 6.0  | —   | —   | 42.5     | 0.015 | —     | 8.22                                       |
| Inconel 718   | 0.04              | 18.6 | —    | 0.4  | 0.9 | 3.1  | —   | 5.0 | 18.5     | —     | —     | 8.19                                       |
| René 41       | 0.09              | 19.0 | 11.0 | 1.5  | 3.1 | 10.0 | —   | —   | 5.0 max. | 0.007 | —     | 8.25                                       |
| Nimonic PK 31 | 0.04              | 20.0 | 14.0 | 0.65 | 2.6 | 4.7  | —   | 5.3 | —        | 0.002 | 0.03  | 8.35                                       |
| Waspaloy      | 0.07              | 19.5 | 13.5 | 1.4  | 3.0 | 4.3  | —   | —   | 2.0      | 0.006 | 0.07  | 8.19                                       |
| G81           | 0.05              | 19.5 | 18.0 | 1.3  | 2.3 | —    | —   | —   | 0.5      | added | —     | 8.18                                       |
| G83           | 0.10              | 10.0 | 10.0 | 2.0  | 3.0 | 10.0 | —   | —   | —        | added | added | —                                          |
| Astroloy      | 0.06              | 15.0 | 15.0 | 4.4  | 3.5 | 5.25 | —   | —   | —        | 0.03  | —     | 7.90                                       |
| René 95       | 0.15              | 14.0 | 8.0  | 3.5  | 2.5 | 3.5  | 3.5 | 3.5 | —        | 0.010 | 0.05  | 8.22                                       |
| EI 437B       | ≤0.06             | 20   | —    | 0.7  | 2.5 | —    | —   | —   | ≤4       | 0.01  | —     | —                                          |
| EI 766        | ≤0.08             | 16   | —    | 0.7  | 2   | —    | —   | 1.3 | ≤3       | —     | —     | —                                          |
| EI 698        | ≤0.08             | 14   | —    | 1.2  | 2.3 | 2    | —   | 1.8 | ≤4       | —     | —     | —                                          |

forging routes has become increasingly evident. Heat treatments have been intensively investigated to develop high-strength properties and low notch sensitivity. The UK initially adopted<sup>125</sup> the more complex Ni–Cr–Co alloys with excellent high-temperature tensile properties, e.g. G81<sup>125</sup> (Nimonic 90), G83,<sup>126</sup> and Nimonic PK 31.<sup>125</sup> The latter alloy possessed a high proof stress up to 550°C and a good creep resistance up to 700°C; its creep and tensile ductility was marginally useful above 600°C and this, coupled with machining problems such as the broaching of blade root fixings that increased component cost, has inhibited its use. The successful USA development of Waspaloy<sup>64, 127</sup> led to its introduction in the UK for HP turbine discs, while Incoloy 901 was adopted for the IP stages. Stronger USA alloys were derived from blade alloy compositions, e.g. René 41 favoured<sup>126</sup> by GE, and more recently Astroloy<sup>128</sup> (a modification of Udimet 700) and René 95.<sup>129</sup>

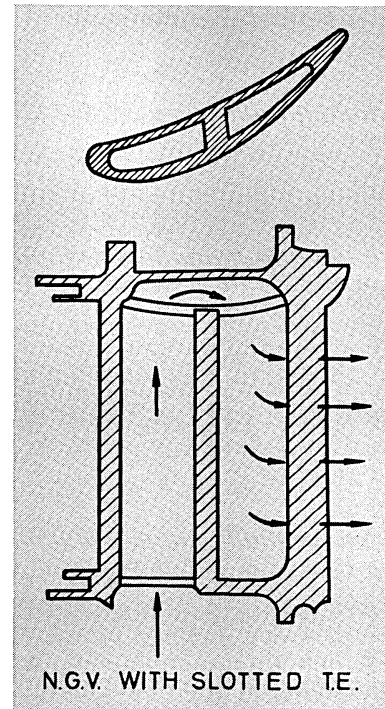
The introduction of improved processes (e.g. rotating electrode and inert gas atomization) for making superalloys commercially in powder form<sup>130</sup> has fostered an interest in powder metallurgy techniques thereby avoiding ingot segregation problems and aiding forgeability, particularly for the strongest disc alloys. Because of its spherical shape and the high strength of the alloy, the powder is consolidated<sup>131</sup> hot by extrusion or by isostatic pressing<sup>132</sup> for larger or complex sections. The fine grain size of the preform permits superplastic deformation during subsequent low-stress isothermal forging. Post-forging heat treatments alter the grain size and precipitate morphology according to the property requirements. Gatorized (*see above*) integral disc-blade components have been made<sup>133</sup> in IN 100 with obvious economies in input weight and machining costs, and have been advocated for supersonic expendable gas turbines for use in missiles. Thermo-mechanical processing<sup>134</sup> is also receiving considerable attention for improving the properties, particularly reducing the scatterband of less advanced alloys. For turbine discs, the USSR has developed both iron- and nickel-base alloys, the earlier compositions being similar<sup>112</sup> to those used in the West, e.g. A286 and Nimonic 80A. Later alloys have no close Western equivalent; EI 698, for example, has<sup>48</sup> somewhat higher titanium plus aluminium content and lower refractory metal (molybdenum and niobium) content than Inconel 718.

#### Nozzle Guide Vane Alloys

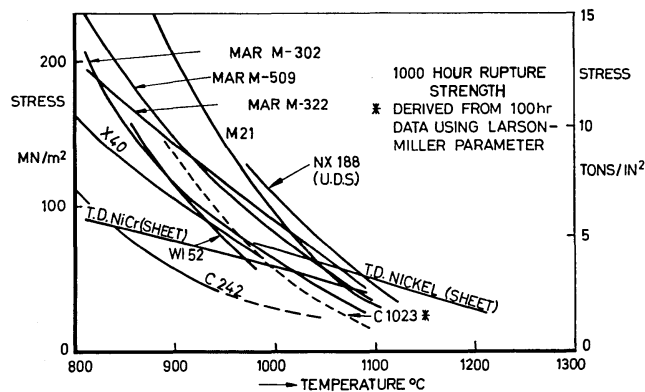
Nickel- and cobalt-base casting alloys are invariably used for cooled and uncooled nozzle guide vanes (sometimes called turbine stator vanes). Although the mechanical loads due to gas bending result typically in stresses of 30–110 MN/m<sup>2</sup>, the

stress requirements for some current designs of HP vane are sufficiently high to necessitate the selection of turbine blade alloys. Failure of these components is denoted<sup>94, 135</sup> by excessive bowing (as a result of the gas-bending loads), local distortion or cracking (by repeated thermal shocks), corrosion or erosion of the leading edge, and occasionally local melting at hot spots created by temporarily erratic combustion. Although internal cooling (*see Fig.17*) and external protection of HP vanes is mandatory to reduce the incidence of local melting and corrosion respectively, more refractory alloys are always under consideration.

Figure 18 indicates the comparative rupture strength merits of some nickel- and cobalt-base vane alloys, their nominal compositions being given in Tables VII and VIII. Nickel-base alloys such as C242 and C1023, and cobalt-base alloys such as



17 Cooled HP nozzle guide vane of Conway engine



18 1000 h rupture stress-temperature for nozzle guide vane alloys

X40 have been extensively used in UK aero-engines but more recently stronger nickel-base alloys with a higher solidus temperature, e.g. PD 21 (M21) have been adopted. The high density<sup>136</sup> of experimental alloys such as the high solidus NASA-WAZ20 alloy containing 20%W may be acceptable for nozzle guide vanes. The improved resistance to thermal fatigue and enhanced life of vanes in directionally solidified MAR-M200 alloy have been demonstrated<sup>106, 138</sup> by engine tests. Although cobalt-base alloys rely largely on carbide strengthening, some systems are hardened by intermetallics, e.g. CoAl in Airesist 213 alloy.<sup>138</sup> As Sims<sup>94</sup> points out, their good structural stability at low stresses makes them particularly valuable for long-life parts. Their good castability, weldability, and thermal fatigue resistance and superior sulphidation resistance to the lower-chromium nickel-base alloys are responsible for their retention, particularly the lower-strength alloys such as FSX 414, in marine and industrial gas turbines.<sup>139</sup>

In relation to conventional superalloys, the high melting point of TD Nickel is superficially attractive for some HP vanes despite its low tensile properties and the necessity of protection. If such vanes were fabricated from sheet, the higher melting surface-alloyed chromium-base and niobium-base alloys (see below) might be preferred. However, the life of

protected components in such candidate materials is much more dependent on the integrity of the coating (e.g. its resistance to foreign object damage) than that in conventional superalloys. The usefulness of high melting point intermetallic compounds has been limited by their generally poor room temperature ductility. The structural stability and strength of Ni<sub>3</sub>Al has merited alloying studies, notably by Maxwell,<sup>140</sup> who showed that a directionally solidified Ni-Mo-Al alloy (NX 188) had potentially attractive rupture strength (see Fig. 18) and ductility for turbine vanes. As in the case of TD Nickel, protection would be essential. Alternatively, transpiration-cooled vanes may be fabricated from porous sheet ('Poroloy') prepared from flattened fine wire space wound in a geometric pattern and diffusion bonded. The cyclic oxidation characteristics of the wire are of major importance for retention of permeability; an Fe-15Cr-4Al-1Y alloy is particularly promising.<sup>141</sup> Laminated porous sheets with an internal grid pattern<sup>142</sup> to provide a high surface area for cooling have been fabricated in Hastelloy X.

### Sheet Alloys

For many years improvements in design of sheet metal components have enabled engineers to rely on relatively simple and easily fabricated solid solution alloys.<sup>143</sup> Nimonic 75 was utilized almost exclusively in UK engines for components such as jet pipes, flame tubes, and exhaust annuli, while Hastelloy X<sup>144</sup> (see Table XI) was favoured for USA engines. Higher operating temperatures created the need for improved sheet nickel alloys, and initially sheet versions of the existing turbine blade alloys, Nimonic 80A and 90, were used. Although possessing improved strength properties, it soon became apparent that their formability and weldability characteristics, at that time ill defined by the user, were inadequate. Closer collaboration between the alloy developer and the user ensued so that the former appreciated the fabrication problems involved and the latter modified design and fabrication techniques in order to minimize constraint during welding and avoid cracking or excessive residual stresses immediately after welding or after post-weld heat treatment. The entire welded structure must be resistant to embrittlement after prolonged exposure under service conditions. The optimum properties in the welded structure, in particular tensile and

TABLE XI  
Nominal Chemical Compositions and Densities of Nickel-Base Alloys for Sheet Applications

| Alloy                 | Compositions, wt-% |      |          | Al       | Ti       | Mo   | W   | Nb          | Fe        | B          | Zr        | Others          | Density, 10 <sup>3</sup> kg/m <sup>3</sup> |
|-----------------------|--------------------|------|----------|----------|----------|------|-----|-------------|-----------|------------|-----------|-----------------|--------------------------------------------|
|                       | C                  | Cr   | Co       |          |          |      |     |             |           |            |           |                 |                                            |
| Nimonic 75            | 0.1                | 19.5 | —        | —        | 0.4      | —    | —   | —           | 5.0 max.  | —          | —         | —               | 8.37                                       |
| C263 (Nimonic 263)    | 0.06               | 20.0 | 20.0     | 0.45     | 2.15     | 5.9  | —   | —           | 0.75 max. | 0.001 max. | 0.02 max. | —               | 8.37                                       |
| PK 33 (Nimonic PK 33) | 0.07 max.          | 18.0 | 14.0     | 2.1      | 2.3      | 7.0  | —   | —           | 1.0 max.  | 0.005 max. | 0.06 max. | —               | 8.21                                       |
| Inconel 625           | 0.05               | 22.0 | —        | 0.4 max. | 0.4 max. | 9.0  | —   | 3.7 (Nb+Ta) | 3.0 max.  | —          | —         | 61 (Ni+Co)      | 8.44                                       |
| Nimonic Alloy PE 11   | 0.1 max.           | 18.0 | 1.0 max. | 0.8      | 2.3      | 5.3  | —   | —           | 46.0      | 0.03       | 0.25 max. | 39 (Ni+Co)      | 8.02                                       |
| Nimonic Alloy PE 16   | 0.1 max.           | 16.5 | 2.0 max. | 1.2      | 1.2      | 3.3  | —   | —           | 35.0      | 0.005 max. | 0.05 max. | 43.5 (Ni+Co)    | 8.02                                       |
| Hastelloy X (PE13)    | 0.10               | 21.8 | 1.5      | —        | —        | 9.0  | 0.6 | —           | 18.5      | —          | —         | —               | 8.23                                       |
| Hastelloy S           | 0.02 max.          | 15.5 | —        | 0.2      | —        | 14.5 | —   | —           | 1.0       | 0.009      | —         | 0.02La          | 8.76                                       |
| Inconel 617           | 0.07               | 22.0 | 12.5     | 1.0      | —        | 9.0  | —   | —           | —         | —          | —         | —               | 8.36                                       |
| IN 586                | 0.05               | 25.0 | —        | —        | —        | 10.0 | —   | —           | —         | —          | —         | 0.015Mg, 0.03Ce | 8.54                                       |

creep ductility and thermal fatigue resistance, must be achieved with the simplest of heat treatments, so that where necessary components can be repair welded satisfactorily at overhaul. The composition of nickel-base alloys in sheet form must therefore be carefully controlled in order to provide<sup>145</sup> the desired balance of strength and ductility before and after fabrication.

Progressive development in the UK has resulted in two well developed precipitation-hardened alloys, C263<sup>146</sup> with a Ti+Al content of 2.5%, and PK 33<sup>78</sup> with a Ti+Al content of 4% (see Table XI). The aging response of such alloys markedly affects their susceptibility to post-weld heat treatment cracking.<sup>147</sup> The titanium and aluminium content has been minimized by using other hardeners, such as niobium, which decrease the age-hardening rate of USA alloys<sup>148</sup> such as Inconel 718 and 625 which are useful to about 700°C, the limit of stability of the Ni<sub>3</sub>Nb phase. Molybdenum is used for solid solution strengthening of alloys such as PK 33 and the USA alloys René 41<sup>149</sup> and Waspaloy. These stronger alloys (see Fig. 19) are more difficult to fabricate, showing a tendency to post-weld heat treatment cracking in complex components, and so the less strong C263 alloy is widely used in British engines. Analogous to the disc applications, nickel-chromium-iron alloys such as PE 11, and PE 16<sup>125</sup> are used in the UK for lower-temperature applications (550°–750°C), e.g. combustion chamber and engine delivery casings.

Recent developments in precipitation-hardened sheet alloys have been concerned with improving oxidation resistance in association with acceptable weldability and stability, e.g. IN 617.<sup>150</sup> Additions of rare-earth elements, e.g. lanthanum and cerium, serve to increase the oxide spalling resistance of the cobalt-based HS 188<sup>151</sup> and the nickel-based alloys Hastelloy S<sup>152</sup> and IN 586,<sup>153</sup> the latter alloy having greater creep strength at 1050°C than established alloys such as Nimonic 75, Hastelloy X, and C263. Parallel with the development of stronger alloys and more complex sheet metal structures, more sophisticated joining techniques such as plasma arc welding<sup>154</sup> have been adopted.

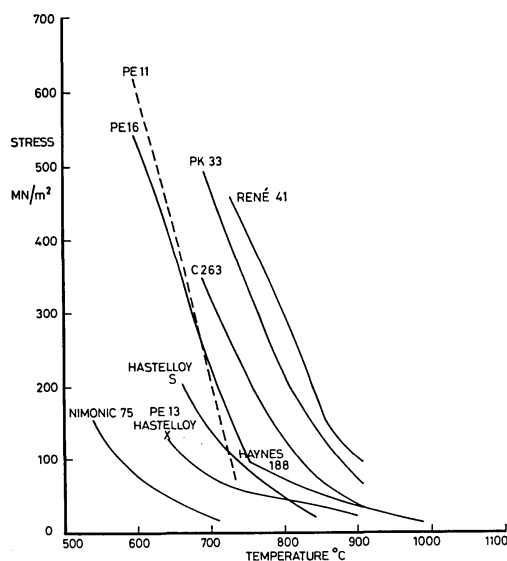
TD Nickel and TD NiCr have been prepared in sheet form but, unlike precipitation-hardened alloys, fusion welding

techniques must not be used, in order to avoid agglomeration of the dispersoid; also, the ductility of these dispersion-strengthened materials diminishes excessively with increase in temperature. It remains to be seen whether these disadvantages and the difficulty of achieving reproducible mechanical properties by controlled thermomechanical processing at an acceptable cost will prevent the application of such alloys in sheet metal structures. The Stellite Division of Cabot Corporation announced<sup>155</sup> the availability of a Ni-16Cr-4Al-1.5Y<sub>2</sub>O<sub>3</sub> alloy in bar form with excellent cyclic oxidation resistance and possessing 10% elongation at fracture in tension at 1093°C. IN 853 (yttria dispersion-strengthened Nimonic 80A) has<sup>156</sup> more attractive low and intermediate strength properties in bar form than the purely dispersion-strengthened alloys, but, equally, alternative joining techniques such as brazing or diffusion bonding appear to be necessary. Sheet versions of both these materials are under development.

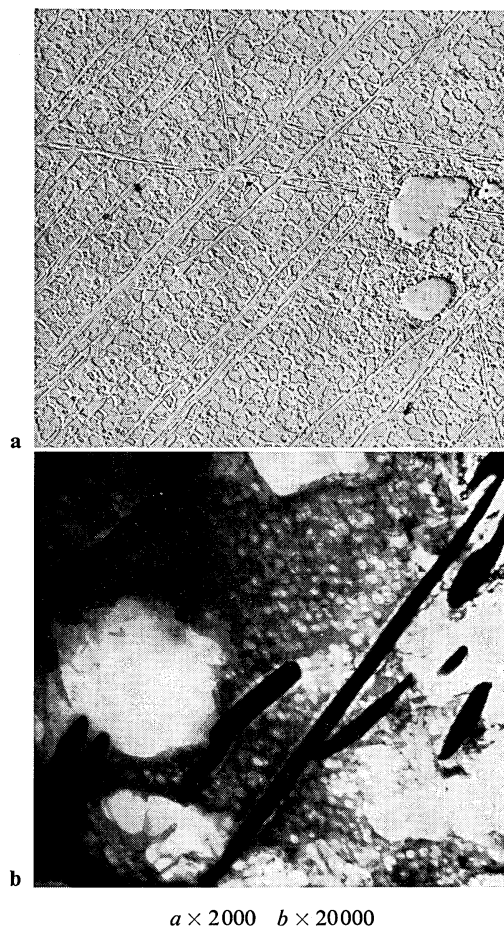
### Structural Stability

Because of the progressive increase in the life of the major structural components of aircraft gas turbines and the growing application of gas turbines to marine and industrial use, the metallurgical stability and corrosion resistance of high-temperature materials are of major importance.<sup>79, 157</sup> Supersonic civil aircraft are designed for  $20 \times 10^3$  h and industrial gas turbines for  $100 \times 10^3$  h. Prolonged exposure of superalloys under stress at elevated temperature induces<sup>158–163</sup> agglomeration of strengthening precipitates, changes in the form and distribution of carbides, formation of intermetallic (often acicular) phases, and alloy depletion at the surface because of oxidation. Early observations<sup>157</sup> showed that linear extrapolation of short-time creep-rupture data on wrought Udimet 700 and cast IN 100 could result in the predicted life being an order of magnitude greater than the actual life. Also, the instability caused by high-temperature exposure could reduce both the strength and ductility at lower temperatures. For example, in Udimet 700 agglomeration of the  $\gamma'$  phase at high temperature, e.g. 982°C, reduced<sup>157</sup> the strength but increased the ductility in tensile tests at 760°C. In René 41, the ratio of the carbide phases  $M_6C$  and  $M_{23}C_6$  markedly decreased<sup>157</sup> with exposure time and temperature; at 760°C, thickening of the grain boundaries by  $M_{23}C_6$  precipitation lowered the ductility, but at higher temperatures the formation of blocky carbides was beneficial to ductility. In order to ensure long-term stability, careful control of composition and microstructure is necessary to avoid the formation of undesirable intermetallic phases during heat treatment or after prolonged exposure to service conditions. Topologically close-packed phases that form from the matrix and deposit<sup>158, 159</sup> along octahedral planes not only embrittle and provide easy paths for crack propagation, but can weaken the alloy because their formation depletes the matrix of strengthening elements, particularly refractory metals. Plate-like phases, analogous to the  $\sigma$  phase in ferritic stainless steels, were identified by Wlodek<sup>160</sup> in the alloy IN 100 which showed zero tensile ductility at 760°C after exposure in a creep test at 870°C and 69 MN/m<sup>2</sup>. The sigma phase in nickel-base alloys nucleates slowly and may take thousands of hours to appear at temperatures between 750° and 950°C (see Fig. 20).

Phase composition (Phacomp) control has now been developed<sup>158, 164</sup> scientifically to avoid the formation of these electron compounds in hot-worked alloys, and to reduce the empiricism in the development of cast alloys. Using the principles of electron-vacancy theory it is possible to predict from the chemical composition of an alloy its tendency to  $\sigma$  formation. The electron-vacancy concentration of the residual



19 0.2% creep strain (100 h) v. temperature for sheet alloys



20 Nature of sigma phase formed in HP Avon turbine blade after prolonged service

matrix after certain carbides and the  $\gamma'$  phase have precipitated is calculated by the linear combinations of elements remaining in the matrix. Hot-worked alloys, for example, which have an electron-vacancy number less than 2.52 are essentially free from  $\sigma$  phase. Woodyatt *et al.*<sup>164</sup> determined the susceptibility of many compositions to  $\sigma$  phase formation, taking into account carbide and boride reactions. Subsequent work<sup>165</sup> on refining the Phacomp technique showed that more precise compositional data on the phases present were necessary. The treatment is less accurate for cast superalloys but has stimulated studies on determining<sup>86, 166</sup> the partitioning of the alloying elements between the three principal phases ( $\gamma'$ , matrix, and carbide) in the microstructure. Despite the assumptions and approximations involved, computerized systems are now regularly used to control production heats, making adjustments in the ratios of critical elements while the alloy is molten in the furnace.

#### Corrosion and Protection

Corrosion has not been a serious problem in the compressors of aero-engines, largely because of the excellent corrosion resistance of 12%Cr steels and titanium alloys, and the efficacy of protective coatings on aluminium alloys. The use of intake separators for hovercraft engines and naval gas turbines has been effective in reducing the salt intake to comparatively innocuous levels, and regular compressor washing<sup>167</sup> or grit

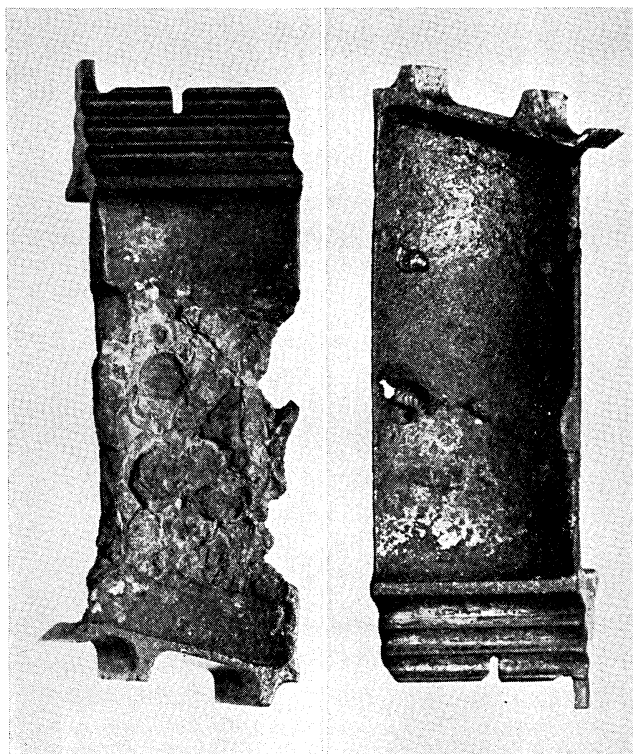
blasting to remove deposited salts has also helped to minimize corrosion. Although hot salt corrosion of titanium alloys has been reported under laboratory conditions, no service problems have been encountered.

A wide selection of low-temperature coatings has been evaluated and introduced<sup>168-170</sup> into marinized engines to reduce sea water corrosion. Anodizing and Nubelon-type paints have been used for aluminium alloys, while cadmium or nickel plating is employed for steel parts. Sermetel W, an inorganically bonded ceramic-metal composite containing aluminium has effectively protected<sup>170</sup> low-alloy steels and magnesium alloys from sea water corrosion. Protection against pitting corrosion, which has led to a drastic reduction in fatigue life of steel compressor blades, is claimed<sup>171</sup> for a coating system (Chromalloy A12) that consists of an inter-metallic compound (formed by pack diffusion), overlaid by a non-metallic layer that suppresses the sacrificial corrosion of the inner layer.

Because of the universal use of distillate fuels of low sulphur and ash content, the structural components of the aircraft gas turbine have been comparatively free from attack by the products of combustion. Resistance to oxidation, i.e. surface stability, of nickel- and cobalt-base alloys is achieved by maintaining a continuous adherent surface scale rich in chromic oxide and/or aluminium oxide. Resistance to nitridation is excellent because of the low nitrogen solubility and low nitride forming tendency of these alloys. Consequently, no serious hot corrosion problems were obtained with alloys such as Nimonic 80 and Nimonic 90 (typically 20%Cr content) in early gas turbines. In order to satisfy the need for increased TET, creep resistance was improved by reduction of the chromium content, thereby allowing a higher content of strengthening alloy additions without seriously impairing oxidation resistance. However, when Nimonic 100 (about 9%Cr) was introduced, premature hot corrosion occurred<sup>172, 173</sup> in the turbine blades of Proteus and Gyron engines. Its replacement by Nimonic 105 (about 15%Cr) was satisfactory in these and other aero-engines, except at longer durations or for operation involving unusual exposure to marine environments, e.g. island-hopping services. Protective coatings became increasingly necessary for this alloy and for stronger cast alloys, particularly in the air-cooled (i.e. thin-walled) HP nozzle guide vanes and turbine blades of marine and power-generating gas turbines of the Tyne and Olympus types. Operational experience with stationary gas turbines is described in Refs. 174 and 175.

Accelerated oxidation, often accompanied by sulphidation, has been particularly evident in the turbine blades of the engines of naval patrol boats and hovercraft where, in the early operations, separators were not employed and salt was ingested into the engine intake; but it has also occurred when sodium chloride has been a contaminant in the fuel. This type of attack, which appears most severe in the temperature range 800°–870°C, is associated with deposits on the blade surface of sodium sulphate, formed by reaction of the ingested salt in the air with sulphur in the fuel during combustion. The catastrophic nature of the attack is evident from Fig.21.

The hot corrosion of gas turbines has been recently and excellently reviewed by Stringer.<sup>176</sup> The mechanisms of initiation and continuation of hot salt corrosion are complex and not thoroughly understood. One simplified explanation is that when the oxide film has been fluxed or rendered discontinuous by chloride contaminants, because of incomplete conversion of the chloride in the sea salt to sulphate during the combustion processes, reduction of the sulphate by the blade alloy or by carbon (as a consequence of erratic combustion) is



21 Salt-corroded turbine blades after service in Proteus engine

followed by diffusion and reaction of the sulphur with the chromium in the alloy, resulting in chromium sulphides and local depletion of the chromium. Thus sulphidation leads to accelerated oxidation. Considerable researches are currently being directed to understanding the problem and to providing metallurgical solutions, i.e. both in improved alloys and protective coating systems. Because of disagreement over the validity of the various laboratory tests used to evaluate the hot corrosion resistance of established and experimental alloys, rig tests in high-velocity gas stream are favoured,<sup>177-180</sup> for reproducing the type of attack found in service. High chromium content is the most effective compositional factor in improving hot corrosion resistance, but because of interaction effects, it is unwise to draw broad conclusions about the independent influence of other alloying elements.<sup>178, 181, 185</sup> A series of wrought alloys, namely M313 (Nimonic 81), IN 587, and IN 597 containing, respectively, 30, 29, and 25%Cr and having superior hot corrosion resistance but correspondingly similar strength to Nimonics 80, 90, and 105 has been developed.<sup>88</sup> Rig tests showed that the cast alloy IN 738 was superior to Nimonics 105 and 115.<sup>181</sup>

It is now generally accepted that the alloying requirements for hot corrosion resistance and high creep strength are so different that protection will always be required for the high-strength alloys. The empirical development and proprietary nature of the coatings have limited the usefulness of several state-of-the-art reviews,<sup>182-184</sup> especially in rationalizing comparative performance assessments. Aluminized coatings have been preferred and although spray, dipping, and pack diffusion processes are used by different engine firms and coating vendors, the performance of such coatings on any one alloy is essentially independent of the process. Performance is affected<sup>186</sup> by substrate composition because of outward

diffusion of alloying elements and inward diffusion of aluminium during the coating process, which influence the ductility of the coating and its subsequent resistance to cracking by thermal fatigue, creep, or foreign object damage in service. Modifications to aluminide (NiAl or CoAl) coatings have been made to upgrade their temperature capability or extend their operating life, e.g. by the introduction of chromium<sup>187, 188</sup> and of barrier layers of tantalum<sup>188</sup> and platinum.<sup>189, 190</sup> However, it is probable that modified aluminide coatings of acceptable ductility are limited to 950°–1000°C. Coatings that do not involve diffusion processes and are tailored to have much slower rates of diffusion in service are being developed<sup>191</sup>; for example, CoCrAlY coating compositions have been applied<sup>192</sup> by electron beam vapour deposition, and showed a 3:1 life advantage over aluminide coatings in burner rig tests.

For the thin-walled HP blades of advanced engines, it will be difficult to develop a coating which is satisfactory over the entire range of temperature experienced by the blade, and so a compromise between resistance to hot corrosion and to oxidation at the maximum blade temperature may be required.

### Chromium-Base Alloys

The good oxidation resistance of chromium and its high melting point and low density in relation to nickel (see Table II) have encouraged researches on chromium-base alloys during the last three decades.<sup>193-195</sup> The development of the iodide process for the production of high-purity chromium provided a strong impetus in the early 1950's for studies of the influence of interstitials and the scavenging and strengthening effects of alloying additions. Like other metals with body-centred cubic structures, chromium exhibits a ductile–brittle transition (DBT) which occurs around 150°C for the commercial purity grade in the recrystallized condition. Unlike nickel, it has a high solubility for nitrogen at high temperatures, and because the oxide film is permeable, the substrate is readily saturated and is embrittled by deposition of the nitride (Cr<sub>2</sub>N) on preferred crystallographic planes. Despite excellent research in Australia, the USA, and to a lesser extent the UK, no usable alloy has emerged.

The high-temperature strength of chromium can be considerably improved by solid solution strengthening with the refractory metals, but this results in a significant increase in the ductile–brittle transition temperature. Strengthening by precipitates, e.g. carbides of titanium, or by dispersions, e.g. magnesia, tends to reduce the transition temperature. In the development of alloys that are competitive with nickel-base alloys, the trade-off between high-temperature strength and low-temperature ductility has involved both strengthening mechanisms. Nitrogen embrittlement during high-temperature air exposure of such alloys cannot be effectively prevented by scavengers such as yttrium and lanthanum, and it has been concluded that surface alloying is necessary. So far, such coatings either are brittle or embrittle the substrate by diffusion, although Cr–Y–La coatings are promising.<sup>195</sup>

The compositions of the main experimental alloys are given in Table XII, and their comparative rupture strengths in Fig. 13. The Australian Department of Supply Laboratories developed<sup>196, 197</sup> three alloys of which only the first, Alloy E (Cr–Ta–Ti), has been extensively evaluated.<sup>198</sup> It has attractive creep–rupture strength and good oxidation resistance but limited ductility, and it is prone to nitrogen embrittlement. Annealing reduced the strength and increased the ductility of the extruded alloy, e.g. the bend transition temperature was reduced from 93° to –73°C. In the recrystallized condition the impact transition temperature was 300°–400°C for unnotched specimens and 500°–600°C for notched specimens, a fact which

TABLE XII  
Nominal Chemical Compositions and Densities of Alloys Based on Chromium and the Refractory Metals

| Alloy     | Composition, wt-%<br>C | Ti   | Mo   | Ni   | Zr   | W    | Hf  | Ta   | V   | Cr   | MgO | Si  | Others     | Density,<br>10 <sup>3</sup> kg/m <sup>3</sup> |
|-----------|------------------------|------|------|------|------|------|-----|------|-----|------|-----|-----|------------|-----------------------------------------------|
| TZM       | 0.025                  | 0.48 | Bal. | —    | 0.09 | —    | —   | —    | —   | —    | —   | —   | —          | 10.22                                         |
| TZC       | 0.14                   | 1.25 | Bal. | —    | 0.12 | —    | —   | —    | —   | —    | —   | —   | —          | 10.05                                         |
| SU 31     | —                      | —    | —    | —    | —    | 17.0 | 2.5 | —    | —   | —    | —   | —   | —          | 9.5                                           |
| AS 30     | 0.1                    | —    | —    | Bal. | 1.0  | 20.0 | —   | —    | —   | —    | —   | —   | —          | 9.60                                          |
| F48       | 0.1                    | —    | 5.0  | Bal. | 1.0  | 15.0 | —   | —    | —   | —    | —   | —   | —          | 9.40                                          |
| B66       | 0.006                  | —    | 5.0  | Bal. | 1.0  | —    | —   | —    | 5.0 | —    | —   | —   | —          | 8.43                                          |
| FS 85     | 0.01 max.              | —    | —    | Bal. | 0.85 | 11.0 | —   | 27.5 | —   | —    | —   | —   | —          | 10.6                                          |
| Alloy E   | 0.01                   | 0.1  | —    | —    | —    | —    | —   | 2.0  | —   | Bal. | —   | 0.5 | —          | 7.32                                          |
| Alloy H   | —                      | —    | —    | —    | —    | —    | —   | 2.0  | —   | Bal. | —   | 0.5 | 0.5RE      | —                                             |
| Alloy J   | —                      | —    | —    | —    | —    | —    | —   | 2.0  | —   | Bal. | —   | 0.5 | —          | —                                             |
| C 207     | 0.1                    | 0.2  | —    | —    | 0.8  | 7.5  | —   | —    | —   | Bal. | —   | —   | 0.15Y      | —                                             |
| CI 41     | 0.09                   | —    | 7.1  | —    | —    | —    | —   | 2.0  | —   | Bal. | —   | —   | 0.1Y(+La)  | —                                             |
| Chrome 30 | —                      | 0.5  | —    | —    | —    | —    | —   | —    | —   | Bal. | 6.0 | —   | —          | 6.36                                          |
| Chrome 90 | —                      | —    | —    | —    | —    | —    | —   | —    | 2.5 | Bal. | 3.0 | 0.5 | —          | —                                             |
| CR 44     | —                      | —    | —    | —    | —    | —    | —   | 2.0  | —   | Bal. | —   | —   | 1.2Y, 0.2B | 7.22                                          |

helps to explain the failure by impact at the roots of Alloy E turbine blades in a rotating rig test and emphasizes the need for ductility at the minimum operating temperature of use. The insensitivity to notches above 600°C was destroyed by prolonged exposure at 1000°C in air. Its thermal fatigue resistance, based on the endurance to the first observed crack, was superior to any nickel-base alloy, but crack propagation rates were very rapid. Recent Australian work showed<sup>199</sup> that cooled nozzle guide vanes could be made in Alloy E by extrusion techniques. In Alloy J the cerium addition produced better retention of ductility after high-temperature air exposure, but difficulty in controlling this addition during manufacture has inhibited development of the alloy. Cascade rig tests up to 1087°C on Alloy J vanes showed<sup>200</sup> that nitrogen contamination was much less in sonic velocity gas streams than in still air; metal losses by oxide formation and spalling were considered acceptable up to about 1050°C.

The USA alloy C207 combines<sup>201</sup> solid solution strengthening (with tungsten) and precipitate strengthening (by carbides of zirconium and titanium) with a yttrium addition for scavenging, and shows a higher rupture strength than Alloy E but a similar impact transition temperature. Even stronger USA alloys, but with higher DBT temperatures, have been developed.<sup>195</sup> A cast Soviet alloy<sup>195</sup> of unknown composition and possessing poor ductility at room temperature has a 100 h rupture strength of about 175 MN/m<sup>2</sup> at 1093°C (cf 48 MN/m<sup>2</sup> for C207 alloy). Relatively little development work in the UK has been published; a Cr-Ta alloy with B and Y additions, designated CR 44, exhibited<sup>202</sup> unacceptably low tensile properties for turbine blades but might be suitable for nozzle guide vanes. The MgO dispersion-strengthened<sup>195</sup> alloys, Chrome 30 and Chrome 90 have low DBTT values but are too weak even for inlet guide vanes. There is no evidence that finer oxide particles, perhaps developed *in situ*, will produce greater strength with acceptable ductility.

The diminished research effort on chromium-base alloys is now more concerned with reducing or preventing nitrogen embrittlement and improving impact ductility than with increasing rupture strength. In view of the present state of development of advanced superalloys and of cooling technology, the applicability of chromium alloys even in short-life engines is unlikely in the foreseeable future.

#### Alloys of the Refractory Metals

The high melting points (see Table II) of the more common

refractory metals,<sup>203</sup> namely niobium, molybdenum, tantalum, and tungsten, indicate the promise of considerably better high-temperature strength properties than nickel-, cobalt-, or chromium-base alloys for turbine blading. Because of the high densities of tantalum and tungsten, only molybdenum and niobium have been seriously examined. Alloys of attractive high-temperature strength<sup>204, 205</sup> have been developed (see Table XII) and are commercially available.<sup>206, 207</sup> Their oxidation resistance, although appreciably higher than that of the respective base metals, is inadequate for turbine blade service. Protective coatings, generally based on silicides, have been developed for service to at least 1200°C. However, such coatings, being thin (typically 0.10 mm) and brittle, are susceptible to impact damage. In the case of molybdenum alloys, coating breakdown<sup>208, 209</sup> has led to localized wastage (as a result of the formation of molybdenum trioxide at temperatures above about 800°C) and often rapid failure because of the increase in centrifugal stress as the effective section area diminished. The catastrophic consequences of coating failure could not be tolerated and so interest in molybdenum alloys ceased in the late 1950's.

Niobium is more attractive<sup>210</sup> than molybdenum because of its lower density, lower DBTT, the refractoriness of its principal oxide, and the greater (but far from adequate) oxidation resistance of its alloys. The high thermal conductivity, low thermal expansion coefficient, and low modulus of niobium alloys, associated with high ductility, implies much higher resistance to thermal fatigue than nickel-base alloys. Niobium has been strengthened<sup>211–213</sup> by the combined mechanisms of solid solution (e.g. with tungsten, tantalum, and hafnium), and dispersion-precipitation (e.g. by hafnium or zirconium carbides). Because the preferred alloys contain substantial amounts of the heavier refractory metals (see Table XII), their densities are as much as 25% greater than those of nickel-base alloys, thus incurring an appreciable weight penalty not only as a result of the heavier blades, but also because of the heavier discs required to accommodate the increased centrifugal loading. Despite the attractive high-temperature strength characteristics of commercially developed alloys such as the UK alloy SU 31,<sup>207</sup> the resulting temperature advantage (about 150°C as indicated in Fig.13) over nickel-base alloys is similar to that obtained by convection cooling of the same alloys. There is no evidence that cooled niobium blades have been seriously considered, but protection of narrow internal cooling passages clearly poses greater problems than that of

external surfaces. The relatively low proof stress of fully heat-treated niobium alloys may necessitate the use of a larger blade root and hence an additional weight penalty for the blade and disc combination. The DBTT of alloys such as SU 31 is below room temperature; protective coatings can raise its value, although the presence of a surface glaze can be beneficial,<sup>214</sup> presumably because of sealing of microcracks in the silicide. Adequate protection under isothermal and cyclic temperature conditions has been demonstrated<sup>214, 215</sup> in the laboratory.

Numerous niobium-base alloys of markedly different composition have been developed in the USA. The strongest alloys, i.e. those rich in tungsten, tantalum, and hafnium, have the highest densities and are difficult to forge, form, and join. Currently there is a diminished interest in applying them to turbine blades. The usefulness of the less strong alloys, e.g. B66 and FS 85, that are available in sheet form is being explored<sup>216</sup> for cooled nozzle guide vanes; for instance, welded B66 constructions, protectively coated with a Si-Cr-Ti slurry and incorporating a nickel-base alloy impingement cooling tube, have been successfully engine tested for 100 h of steady and cyclic state operation at metal temperatures above 1093°C. Coated burner assemblies in FS 85 alloy are scheduled for rig and engine evaluation. Nevertheless, it is felt that clad coatings on the more ductile niobium alloys must replace the damage-sensitive silicide diffusion coatings, particularly where thin sections are involved.

As niobium alloys are several times more expensive than nickel-base alloys, their eventual use may be confined to critical and costly components for which a marked temperature advantage can be clearly defined and demonstrated, e.g. in short-life military engines, or in specialized power plant, e.g. in gas turbines for circulating helium in gas-cooled reactors.

### Ceramics

Bulk ceramics do not appear in production aero gas turbine components. However, the increasing temperature of the delivery air from aero-engine compressors and from the heat exchangers of automotive and truck gas turbines has reduced the efficacy of cooling, particularly in combustion chambers, and so the usefulness of low-expansion (i.e. thermal shock-resistant) refractory materials such as silicon nitride<sup>217</sup> is currently being re-examined in both the USA<sup>218-219</sup> and the

UK.<sup>220-221</sup> Its attractively low density and high melting point (see Table II), low thermal expansion coefficient (about  $2.5 \times 10^{-6}/^{\circ}\text{C}$ ), excellent oxidation resistance, and good strength retention at high temperature (see Fig. 13) are counterbalanced by a complete absence of ductility and therefore by poor resistance to crack propagation and to mechanical shock. Although the lack of plastic deformation in association with its low thermal expansion characteristics are virtues as far as dimensional stability is concerned, local peak stresses cannot be relieved. As in the case of all brittle materials, appreciable scatter in strength properties is evident and so, for material acceptance, statistical evaluation of the relevant mechanical properties is essential. Proof testing of the more highly stressed components<sup>222, 223</sup> will almost certainly be required. Nevertheless, silicon nitride is particularly attractive for low-cost, efficient, lightweight power units.<sup>224</sup> The creep characteristics of hot-pressed grades indicate that service temperatures close to 1200°C and turbine entry temperatures as high as 1370°C are possible. Applicability in combustion chambers and especially in the turbine will depend on their mechanical reliability and the extent to which impact damage can be avoided. Silicon carbide can be even more creep resistant<sup>221</sup> than hot-pressed silicon nitride, but its higher thermal expansion coefficient (about  $4.0 \times 10^{-6}/^{\circ}\text{C}$ ) may render it more sensitive to severe thermal shock.

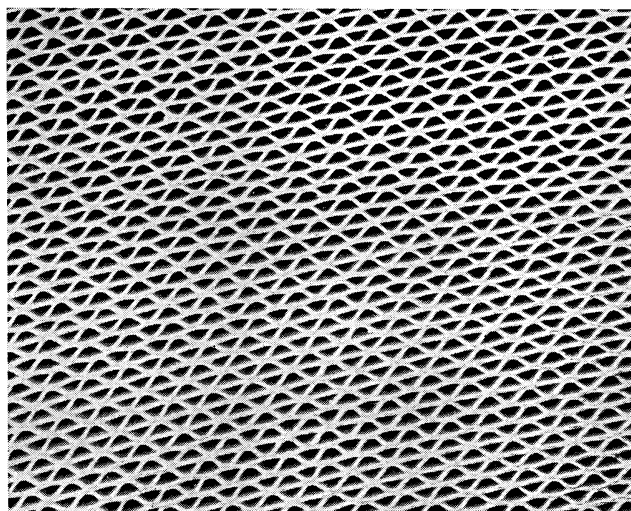
Rotary recuperative heat exchangers of ceramic construction have been proposed<sup>225</sup> for road transport gas turbines, and lives up to 4000 h in demonstration engines have been achieved.<sup>16</sup> The matrix shown in Fig. 22 is supported between inner and outer rings of the same composition, e.g. lithia-alumino-silicate produced by the heat treatment of a nucleated glass. The matrix material density is about 20% that of stainless steel and, as its coefficient of expansion between 0 and 800°C is only about  $2 \times 10^{-6}/^{\circ}\text{C}$ , thermal stresses are low and gas seal clearances can be maintained.

### Fibre Composites

Since the futile attempts in the 1950's to develop ceramic-metal mixtures (cermets) for turbine blading, resulting<sup>226</sup> in near-brittle composites with little strength advantage over nickel-base alloys, fibrous rather than particulate forms of non-metals have been and are being exploited.

Fibre reinforcement of organic or metallic materials is attractive because the resulting composites can be lighter, stiffer, and stronger (see Table XIII) than established alloys. In lift jet engines, e.g. RB 162, where weight is all-important, glass-reinforced plastics have been applied extensively<sup>7, 227</sup> because of their low density and adequate specific strength. For more advanced lift engines and for the front fan and early stages of large subsonic engines, stiffer fibres such as carbon<sup>227-229</sup> and boron<sup>230, 231</sup> are potentially useful for reinforcing resins<sup>232</sup> and aluminium alloys<sup>233</sup> respectively, thereby providing composites with lower density than and superior stiffness to titanium alloys. For turbine blading, the stability of the fibre in a cobalt- or nickel-base alloy matrix is of paramount importance and severely restricts the choice of useful reinforcements. Ignoring cost considerations, the applicability of all such materials is strongly influenced by their behaviour in engine environments involving corrosion and erosion/impact damage.

Considerable research and development is in progress on low-temperature composites using carbon fibres produced commercially in the UK and boron and carbon fibres in the USA. Carbon fibres with a highly oriented graphite structure are produced<sup>234-236</sup> commercially by heat treatment of polyacrylonitrile fibre, involving stretching and oxidation at



22 View of section through Cercor heat exchanger matrix

TABLE XIII  
Comparative Data on Compressor Blade Materials

| Material                   | Density $\rho$ ,<br>$10^3 \text{ kg/m}^3$ | UTS $\sigma$ ,<br>$\text{MN/m}^2$ | Flexural<br>modulus $E$ ,<br>$\text{GN/m}^2$ | $\frac{\sigma}{\rho}$ , $\text{MNm/kg}$ | $\frac{E}{\rho}$ , $10^3 \text{ GNm/kg}$ |
|----------------------------|-------------------------------------------|-----------------------------------|----------------------------------------------|-----------------------------------------|------------------------------------------|
| Al alloy                   | 2.71                                      | 420                               | 69                                           | 0.155                                   | 25.5                                     |
| Ti alloy                   | 4.51                                      | 1004                              | 124                                          | 0.222                                   | 27.5                                     |
| Steel                      | 7.75                                      | 1080                              | 207                                          | 0.139                                   | 26.7                                     |
| Glass fibre resin          | 2.05                                      | 1130                              | 48                                           | 0.550                                   | 23.4                                     |
| Carbon fibre resin (Hyfil) | 1.60                                      | 1160                              | 165                                          | 0.725                                   | 103.1                                    |

low temperature of the polymer, followed by pyrolysis in the range  $1000^\circ\text{--}2500^\circ\text{C}$ , typically giving tows of 10000 filaments, each  $6 \mu\text{m}$  in diameter. For example, Type I carbon fibre, resulting from high-temperature pyrolysis has a high modulus while Type II has a lower modulus but higher strength (see Table XIV). After impregnating the spread tows with resin, components such as RB 211 fan blades<sup>237</sup> have been prepared by laying-up laminations and pressing to final shape in heated moulds, followed by polymerization of the resin. The properties can be varied by surface treatment of the fibre and by altering the relative orientation of the laminations, e.g. to produce greater torsional stiffness, thereby permitting some latitude in 'designing' the material to suit different requirements. The advantage in room temperature specific strength and stiffness of Hyfil (carbon fibre reinforced epoxy resin manufactured by Rolls-Royce) over current compressor blade materials is evident from the data in Table XIII. The merits of Hyfil for compressor fan blading are described in Refs. 232 and 237. Although limited flight trials of Hyfil blades in Conway engines were encouraging,<sup>232</sup> insufficient development time was available to improve resistance to bird impact damage and so Hyfil was replaced<sup>238</sup> by Ti-6Al-4V alloy for the RB 211 fan blades.

Boron fibres typically  $100 \mu\text{m}$  in diameter, were the first high modulus fibres to be developed and were prepared<sup>234</sup> by deposition from the vapour phase (i.e. hydrogen reduction of boron trichloride) on to a tungsten wire substrate. Comparative density and property data (see Table XIV) indicate that the specific strength of boron is similar to Type I and its specific modulus intermediate to Types I and II carbon. Boron fibres are relatively heavy, but this disadvantage might be reduced by use of a lower-density substrate such as silica or carbon.<sup>234</sup> However, it is generally accepted that the large-scale production cost of boron fibres is likely to be appreciably greater (perhaps by an order of magnitude) than that of carbon.

For higher-temperature service, e.g.  $150^\circ\text{--}500^\circ\text{C}$ , the usefulness of polyimide-type resins, aluminium, titanium, and nickel matrices is being examined. The manufacture of diffusion-bonded boron-aluminium composites has advanced<sup>239</sup> to the stage where second-stage fan blades have been made by lay-up techniques using laminations prepared by filament winding on to formers covered with aluminium foil and coating the exposed fibres with aluminium deposited by flame spraying. Silicon carbide filaments offer greater stability in aluminium, but, on economic grounds, silicon carbide coated boron (Borsic) is preferred and has been produced commercially.<sup>240</sup> Some properties of 50 wt-% Borsic-aluminium are given in Table XIV. Such composites offer a 40% weight reduction but their impact strength is similar to that of cast aluminium alloys and considerably less than that of forged Ti-6Al-4V alloy. In contrast, carbon-aluminium composites are potentially cheaper but have proved more difficult to

process and have exhibited disappointingly low impact and transverse strength properties.<sup>241, 242</sup> They are prone to corrosion because the carbon-aluminium system has a strong corrosion potential.<sup>242</sup> Engine tests for durations up to 564 h and involving cyclic operations have been successfully reported on Borsic-aluminium alloy fan blade assemblies.<sup>243</sup> It is claimed that silicon carbide whiskers can be produced more cheaply than silicon carbide filaments; sorting and alignment techniques have been developed.<sup>244</sup> Cheaper fibres such as steel wires cannot be used in sufficiently large volume fractions for significant strengthening because of their high density; beryllium wires are both too costly and too low in strength to provide viable composites. Relatively little information is available on boron or silicon carbide reinforced titanium alloys.<sup>245</sup>

For composites intended for use above  $1000^\circ\text{C}$ , the requirement that the fibre must be stable in an oxidation-resistant matrix during processing and subsequent service has severely limited the choice of refractory fibre. Of those that are sufficiently strong to warrant consideration, only tungsten and alumina are attractively stable.<sup>246-248</sup> Composites consisting of a dispersion of tungsten-rhenium<sup>247</sup> or tungsten-thoria<sup>248</sup> wires at 20-70% by volume concentration in a cast nickel alloy matrix offered up to  $90^\circ\text{C}$  temperature advantage over the unreinforced superalloy (see Fig.16). For turbine blades, their high density and therefore weight penalty is apparently less discouraging to USA than to UK workers. Alumina fibre composites can be prepared in a similar way but such fibres (single-crystal sapphire) are very expensive, and because of their brittleness they break up in the matrix under thermal cycling conditions.<sup>247</sup> Alumina whisker composites have been investigated but interaction between whisker impurities and the nickel matrix during processing considerably weakened<sup>249</sup> the reinforcing phase.

There is considerable current interest<sup>250</sup> in aligned two-phase materials in which the reinforcing phase is produced within the matrix, i.e. *in situ* composites, in contrast to those

TABLE XIV  
Properties of Fibres and Low-Temperature Composites

| Material                                 | Density,<br>$10^3 \text{ kg/m}^3$ | UTS,<br>$\text{MN/m}^2$ | Flexural<br>modulus,<br>$\text{GN/m}^2$ |
|------------------------------------------|-----------------------------------|-------------------------|-----------------------------------------|
| Boron filament ( $100 \mu\text{m}$ )     | 2.56                              | 3300                    | 400                                     |
| Type I carbon fibre ( $8 \mu\text{m}$ )  | 1.92                              | 2580                    | 335                                     |
| Type II carbon fibre ( $8 \mu\text{m}$ ) | 1.77                              | 2940                    | 235                                     |
| 50% B in Al (6061-T6)                    | 2.63                              | 1200                    | 207                                     |
| 50% B in epoxy resin                     | 1.90                              | 1730                    | 210                                     |
| 55-60% Type I carbon in epoxy            | ~1.60                             | 1250                    | 210                                     |
| 55-60% Type I carbon in epoxy            | ~1.51                             | 1550                    | 135                                     |

in which the matrix is 'added' to preformed fibres. Directional solidification of eutectic or near-eutectic compositions has produced fine, dendrite-free structures of binary alloys and of the more technologically promising ternary and higher order systems. The load-carrying phase may be fibrous or lamellar; fibres of high aspect ratio and whisker-like strength have been isolated from their eutectics. The most promising aligned eutectics so far developed are reinforced with TaC fibres<sup>251</sup> and Ni<sub>3</sub>Nb lamellae<sup>252</sup> respectively. Bibring *et al.*<sup>251</sup> have 'designed' both cobalt- and nickel-base alloys with well aligned TaC fibres, the nickel alloys being the more attractive because their high  $\gamma'$  content provides strengthening at intermediate temperatures. They also possess good impact and high-cycle fatigue strength and oxidation resistance. The Ni<sub>3</sub>Al-Ni<sub>3</sub>Nb ( $\gamma'-\delta$ ) lamellar<sup>252</sup> eutectic is stronger, showing some 70°C advantage over directionally solidified superalloys such as MAR-M246 (see Fig.16), but has limited oxidation resistance and cold ductility. Blade shapes have been prepared.<sup>253</sup> Despite natural limitations on the volume fraction, morphology, and composition of suitable phases, and some reservations about the stability of *in situ* composites under thermal cycling conditions,<sup>254</sup> there is considerable potential in this class of material.

### Trends and Prospects

The selection of materials for the critical components of gas turbines increasingly requires more sophisticated evaluations and judgments. For example, isothermal fatigue tests with a simple saw tooth cycle are no longer adequate for assessing the low-cycle fatigue behaviour of turbine or compressor discs. Even on established materials, there is a need for knowledge of the effects on crack growth rate of cyclic stresses and temperatures with appropriate dwell times (in order to determine creep-fatigue interactions) and of superimposed vibratory (high-frequency) stresses. If the trend towards consideration of lower ductility and fully brittle materials persists, statistical evaluation of mechanical properties will be essential. Better definition of geometry-dependent characteristics such as oxidation and/or corrosion resistance, weldability, and castability will be required in order to reduce the need for intuitive judgments. Reappraisal of requirements and of the procedures for 'trading-off' properties will involve even closer collaboration between the designer and the materials scientist/engineer. Although more sophisticated treatments, e.g. the application of fracture mechanics will be increasingly involved in laboratory assessments, validation of aero-engine components by rig, engine, and flight testing will always be necessary to demonstrate that both the material and the manufacturing route to the component have been correctly chosen. The area of uncertainty about any new material must be reduced to the minimum before commitment, in order to avoid excessive development in service or costly substitution at a later stage. This applies particularly to large industrial power generating gas turbines which, as Cockcroft<sup>15</sup> points out, are sold from the drawing-board with little post-design development.

With few exceptions (e.g. fibre composites) materials innovations during the past ten years have been less noticeable than the development of processing techniques leading to a technically superior or a lower-cost component.<sup>255</sup> This is evident from the increased emphasis on casting techniques (e.g. directional solidification), electron beam welding,<sup>58</sup> diffusion bonding (e.g. of protective coatings and of hollow titanium fan blades<sup>256</sup>), powder metallurgy techniques to aid forgeability, and electrochemical and electrical discharge machining processes. In the 1960's, progressive advances in cooling technology reduced the need for radically new materials.

Rolls-Royce engines now in service with air-cooled turbine blades operate<sup>257</sup> at combustion outlet temperatures between 1450 and 1650 K. With increase in the temperature of compressor delivery air, the effectiveness of air cooling is now diminishing and the need for strong, refractory materials is becoming more urgent. Past experience has shown that the potential of many such materials has not been realized because of unacceptable ductility or oxidation resistance. Although more ingenious techniques of cooling, possibly using fluids other than air, may be developed, the challenge to materials scientists and engineers and to the designer of gas turbines is a formidable one.

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## Appendix

| Alloy Designation or Prefix | Originator of Trade Name                                                          |
|-----------------------------|-----------------------------------------------------------------------------------|
| AF2 1DA                     | Universal Cyclops Steel Corp., Philadelphia, USA                                  |
| A 286 (Tinidur)             | Fried. Krupp Steel Works, Essen, Germany                                          |
| Airesist                    | Airesist Manufacturing Company of Arizona, Division of Garrett Corp., USA         |
| Borsic                      | Pratt and Whitney Aircraft, Division of United Aircraft Corp., East Hartford, USA |
| B66                         | Westinghouse Electric Corp., Wheeling, USA                                        |
| C242, C263, C1023           | Rolls-Royce (1971) Ltd, Derby, England                                            |
| Chrome 30, Chrome 90        | The Bendix Corp., Southfield, Michigan, USA                                       |
| Chromalloy                  | Chromalloy Corp., White Plains, New York, USA                                     |
| FS 85                       | Fansteel Metallurgical Corp. Inc., North Chicago, USA                             |
| FSX 414, FSX 418            | General Electric Co. of America, Schenectady, USA                                 |
| FV 448, 458, 520B, 535      | Firth Vickers Stainless Steels Ltd., Sheffield, England                           |
| G18B, 64, 81, 83            | Jessop-Saville Ltd., Sheffield, England                                           |
| GMR 235                     | General Motors Corp., Dayton, USA                                                 |
| Grafil                      | Courtaulds Ltd, Coventry, England                                                 |
| H46, 53                     | Jessop-Saville Ltd, Sheffield, England                                            |
| Hastelloy                   | Haynes Stellite Division of Cabot Corp., Kokomo, USA                              |
| Haynes (HS)                 |                                                                                   |
| Hecla 153                   |                                                                                   |
| HTG 3                       | Hadfields Ltd, Sheffield, England                                                 |

|            |                                                                                                                                           |
|------------|-------------------------------------------------------------------------------------------------------------------------------------------|
| Hiduminium | High Duty Alloys Ltd, Slough, England                                                                                                     |
| Hyfil      | Rolls-Royce (1971) Ltd, Derby, England                                                                                                    |
| Hylite     | Jessop-Saville Ltd, Sheffield, England                                                                                                    |
| I 400      | Brush Beryllium Corp., Cleveland, USA                                                                                                     |
| IMI        | Imperial Metal Industries (Kynoch) Ltd, Birmingham, England                                                                               |
| Inconel    | H. Wiggin and Co. Ltd, Hereford, England, and Huntington Alloy Products Division of International Nickel Company Inc., West Virginia, USA |
| Incoloy    |                                                                                                                                           |
| IN         | International Nickel Ltd, London, England                                                                                                 |
| Jethete    | United Steel Companies Ltd, Sheffield, England                                                                                            |
| Lockalloy  | Lockheed Aircraft Corp., Burbank, California, USA                                                                                         |
| M21, M22   | International Nickel Ltd, London, England                                                                                                 |
| M252       | General Electric Co. of America, Schenectady, USA                                                                                         |
| MAR-M      | Martin Marietta Corp., Wheeling, USA                                                                                                      |
| Modmor     | Morganite Modmor Ltd, London, England                                                                                                     |
| MRS-B      | Magnesium Elektron Ltd, Manchester, England                                                                                               |
| Nimonic    | H. Wiggins and Co. Ltd, Hereford, England                                                                                                 |
| Nimocast   |                                                                                                                                           |
| Poroloy    | The Bendix Corp., Filter Division, Michigan, USA                                                                                          |
| René       | General Electric Co. of America, Schenectady, USA                                                                                         |
| René 41    | Allvac Metals Co.                                                                                                                         |
| RZ 5       | Magnesium Elektron Ltd, Manchester, England                                                                                               |
| S816       | Allegheny Ludlum Steel Corp., Philadelphia, USA                                                                                           |
| SU 31      | Imperial Metal Industries (Kynoch) Ltd, Birmingham, England                                                                               |
| TMCA       | Titanium Metals Corp. of America, New York, USA                                                                                           |
| TZC        | Climax Molybdenum Co., New York, USA                                                                                                      |
| TZM        |                                                                                                                                           |
| Udimet     | Special Metals Inc., New Hartford, USA                                                                                                    |
| Vitallium  | Austenal Inc., Division of Howmet Corp., New York, USA                                                                                    |
| Waspaloy   | Pratt and Whitney Aircraft Division of United Aircraft Corp., East Hartford, USA                                                          |
| WI 52      | Waimet Alloys Co., Dearborn, USA (patented by the Tungsten Institute)                                                                     |
| ZE 63      |                                                                                                                                           |
| ZT 1       |                                                                                                                                           |
| ZW 6       | Magnesium Elektron Ltd, Manchester, England                                                                                               |