HEAT RESISTANT MATERIALS: A SURVEY OF SOME BRITISH DEVELOPMENTS

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Summary—The principal developments in Great Britain during the past fifteen years regarding heat- and creep-resisting materials for aero gas-turbines are outlined. Sections deal with austenitic and ferritic alloy steels for disks, and the major types are described. The wide-spread use of the 12% chromium multi-alloy type disk-steel is pointed out, and reasons for its current pre-eminence are given. Stainless steel and nickel-base alloys for flame-tubes and inlet-nozzle guide-vanes are next described, with their main performance requirements.

The evolution of turbine-blade materials is traced, and the current supremacy of the nickel-base "Nimonic" series is noted. The possibilities of certain alternatives are pointed out, especially when turbine-blade working-temperatures exceed 900°C. Cognisance is taken of the influence of some newer techniques, such as hot-extrusion and vacuum-melting and casting; the former being first pioneered on a production scale in France, and the latter in the United States.

British work in the field of powder-metallurgy on the fabrication of both solid and ducted components for cooled gas-turbines has been summarized. Advances in deep-hole spark-drilling are noted in relation to orthodox solid forgings, powder-metal compacts, extrusions and precision-cast components. The prospects of transpiration and sweat cooling are discussed in the light of current progress in materials, rather than of engineering design. Some speculations on possible future lines of advance conclude the paper.

1. INTRODUCTION

When the gas turbine was first being developed during the period 1938–1943, there was considerable mystery attached to the choice of suitable heat-resisting materials for turbine-disks, turbine-blades, inlet-nozzle guide-vanes and flame-tubes. The early designers selected the best of the then available materials, but quite quickly new and specially developed materials started to appear, and have been continuing to do so ever since. There is now a wealth of published literature, and even a range of textbooks dealing with these matters.

The object of this paper is to furnish a critical outline of the major British achievements in this general field during the past fifteen years, with particular emphasis on those materials which have been fully proved by research and verified under gas-turbine service conditions. The many alternatives enforce brevity, but even so this should serve to high-light the salient features and ensure continuity of narrative.

In Great Britain, the first turbine disks were forged from 20 : 8 chromium–nickel austenitic steels having 1·2% titanium and a carbon content of the order of 0·2%. Steels of this type (Firth Vickers: "Stayblade") (Table 1) were forged readily, but were machined with difficulty; particularly as regards the broaching of the fir-tree roots in the turbine-blade slots. They were severely limited in maximum rim operating temperature (about 550°C) and in mechanical proof-stress properties at ordinary temperatures. The turbine-blades were forged and machined from a similar steel at first (or F.V. Rex 78), while nozzle-blades were made from high nickel–chromium steels. Flame-tubes were then being fabricated from various stainless-steels in sheet form.

In mid-1943, experimental jet-engines were failing on the test-bed at the rate of about one per week, and the demand for new high-strength materials was insistent. The early dramatic history has been excellently recorded by Sir Frank Whittle(1). The materials described in this paper succeeded those in use at that particular stage.

Towards the end of 1943, Sir Winston Churchill decided that the jet-engine, installed in fighter planes, might still be needed for victory in World War II, and thus official help and priority were given to all organizations contributing to that end. The National Physical Laboratory assisted magnificently at that time by undertaking creep-tests, hot-fatigue tests, and other investigations to ascertain the performance of hurriedly-developed new steels and alloys. Individual firms concentrated on producing alloys to meet specific requirements, while the engine-builders endeavoured to prove the selected material in complete power units. Rapid progress resulted, and a new and growing industry had been launched. The specific applications will now be discussed and the principal alternatives described.

2. STEELS AND ALLOYS FOR GAS-TURBINE DISKS

The first notable disk steel which appeared at the time when experimental rotors were failing on the test-beds, was a high carbon (0·45%), 13 : 13 : 10 nickel–chromium–cobalt alloy steel containing tungsten, and the (then) novelty of an appreciable addition of niobium(2,3). This steel (Jessop G.18B), developed from the austenitic valve steel DTD49B, dominated the gas-turbine field for several years, both for disks and later for large welded land-based gas-turbine rotors(3,4) (e.g. Sulzer: Weinfelden, Switzerland). It emerged from attempts to free the valve steel (Jessop G2), used in the famous Rolls-Royce "Merlin" engines, from long-term embrittlement; when it was heated to 550°C for several hundred hours. Additions of titanium cured this embrittlement, but niobium was much to be preferred. Experiments with comparatively large additions of niobium(2) (a few per cent) rapidly conferred extraordinary properties on the otherwise well-known steel. Many thousands of G.18B turbine disks have been made and their performance record has been one of outstanding reliability. Johnson(5) has recorded that between 1943 and 1949, 99% of jet-engine austenitic turbine disks in Great Britain were made in G.18B.

The first application of G.18B disks involved a major exercise in flash-butt welding(6), to enable an integral disk and shaft to be produced for replacements in an existing turbine. Even these stringent requirements were fully met, and the resulting jet-engines were the first to fly in fighters against the "Flying Bomb" at the height of the attack on Southern England in late 1944. G.18B steel is fully austenitic when solution-treated, but its mechanical proof properties are markedly improved by "warm-working"(2) (also called "hot-cold" working). Many British gas-turbine engines have disks supplied in the "warm-worked" condition(7). This steel segregates to only a small extent on cooling from the liquid state; and this feature, coupled with considerable ductility at both low and elevated temperatures, made available an excellent material for single-stage jet-engines operating with rim-temperatures up to at least 650°C. "Warm-working" overcame the tendency for undue radial stretching to occur during speeding-up from the cold. Disks of nearly 20 in. diameter were held to within 0·002 in. on diameter for many hundreds of hours of service. Under tensile stress, initial creep in solution-treated G.18B was large, but the creep-strain rapidly diminished as soon as precipitation and internal transformations were promoted by deformation. The "primary creep" was minimized by "ageing-treatments" or by "warm-working". "Warm-working" possessed the advantage of maintaining good creep properties while enhancing the low temperature proof-stress values. Its long-term creep properties are also fully established, and results for as long as 80,000 hr have now been obtained (see Fig. 1). These data currently interest power plant designers.

When multi-stage turbo-jet engines were developed, disk-cooling by

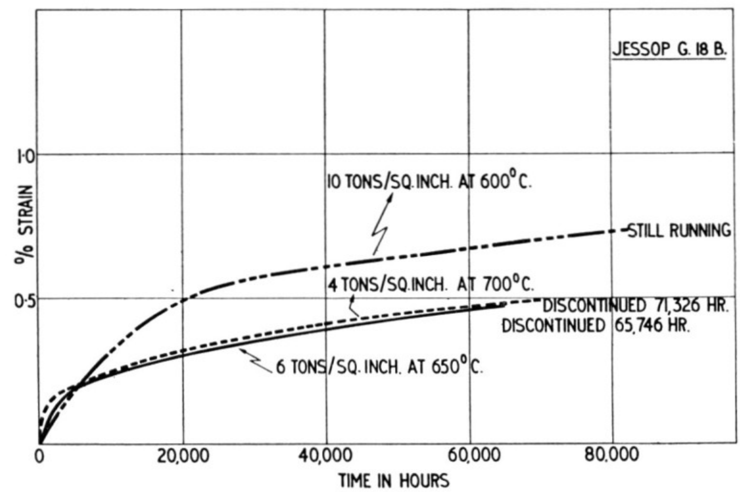

FIG. 1. Long-time creep curves on Jessop G.18B steel for up to 80,000 hr duration.

TABLE 1

Chemical constitution of various heat-resisting high-creep-strength steels and alloys

Alloy	Composition—weight per cent												
	C	Ni	Cr	Co	Mo	W	V	Nb	Fe	Ti	Cu	Ta	Al
(F.V.) Stayblade	0·22	8·5	20	—	—	—	—	—	Balance	1·2	—	—	—
(F.V.) Rex 78	0·1	18	14	—	4	—	—	—	Balance	0·6	4·0	—	—
(J) G.18B	0·4	13	13	10	2	2·5	—	3·0	Balance	—	—	—	—
(J) G.32	0·3	12	19	45	2	—	2·8	1·3	Balance	—	—	—	—
(J) G.42B	0·25	15	19	25	3·0	3·0	—	—	Balance	—	—	—	—
(J) G.56	0·05	25	13	—	2·8	—	—	0·4	Balance	1·8	—	0·4	0·2
(J) G.34	0·8	12	19	45	2	—	2·8	1·3	Balance	—	—	—	—
(J) H.40	0·2	0·4	2·8	—	0·5	0·5	0·75	—	Balance	—	—	—	—
(J) H.46	0·15	0·5	12	—	0·5	—	0·3	0·3	Balance	—	—	—	—
(J) H.53	0·1	—	10·5	7	0·9	0·8	0·5	0·4	Balance	—	—	—	—
(F.V.) 448	0·12	0·7	10·5	—	0·75	—	0·15	0·45	Balance	—	—	—	—
(F.V.) 535	0·1	—	10·5	6	0·75	—	0·15	0 45	Balance	—	—	—	—
(U.S.) Discalloy	0·05	25	13	—	2·8	—	—	—	Balance	1·8	—	—	0·2
(U.S.) A.286	0·05	26	15	—	1·8	—	0·3	—	Balance	2·0	—	—	0·2

radial air-flow associated with more precise inlet-nozzle guide-vane design, enabled turbine-blade root-temperatures to drop somewhat, even though peak turbine-blade temperatures rose. The demand therefore declined for disk materials having still greater creep strengths at temperatures exceeding 650°C. In fact, in some cases rim-temperatures dropped by at least 100°C, and this enabled 12% chromium stainless steels to be used for second- and third-stage disks.

The next phase of development was an exploration of whether turbine-disks having high tensile and proof stresses could be produced in the so-called "ferritic steels" and simultaneously meet the creep-strength requirements. In these, the metallurgical structure is generally that of "tempered martensite", which is usually produced by oil-quenching the steel from above the critical point, followed by one or more tempering operations. It was anticipated that, because the coefficients of thermal expansion were lower, and the thermal conductivities were higher, for ferritic steels; that the thermal-stress effects in the disk on heating-up and cooling-down would be beneficially reduced.

The late Major Halford designed the De Havilland series of single-stage jet-engines, right from the start, using ferritic disks; and a range of low-alloy steel disks was pioneered in his turbines with considerable success. However, as new turbine requirements arose, so did the demand for better properties; and, here it is opportune to pay tribute to a steel developed first in Germany and produced originally for the oil-cracking industry (I.G.F.N10). It had to be particularly resistant to hydrogen-embrittlement. A low-alloy type containing 3% Cr, ½%W, ½% Mo but having the unusually high vanadium content of 0·75% was found to meet the need. During World War II such a steel was used in some of the German jet-engines (JUNKERS JUMO 004). The first of the high-vanadium ferritic steels[3,8] (Jessop H.40) based on the German discovery was then made available to the British aircraft industry. This steel suffered from a limited amount of scaling trouble, as under turbine-operating conditions the disk material was subjected to oxidation at the maximum rim-operating-temperature of 600°C, even though the proof-properties and the creep-strengths at that temperature were more than adequate. At that time (1947) G. T. Harris concentrated on exploring the possibility of producing a stainless steel of the martensitic type, which would have the scale-resisting properties of an orthodox 12% chromium steel, while at the same time retaining the outstanding creep-properties of the H.40 high-vanadium low-alloy steel. Adding chromium to a low-alloy steel usually lowers the creep-resisting properties, and therefore it was some time before the first successful tempered-martensite, high-strength and stainless disk-steel was evolved[3] (Jessop H.46) which is now made under licence in many countries. Its employment facilitates the design of non-scaling lightweight disks of adequate strength and performance. In 1956, Johnson[5] wrote "most modern engines use H.46 type of material, i.e. 12% CrMoVNb steel to the exclusion of non-stainless ferritic types, and austenitic steels such as G.18B. The latter is still in use for some of the older engines, but is rapidly becoming obsolete in favour of the stainless ferritic disk, which is being used in practically all new engines". A steel having very similar composition and mechanical properties is Firth-Vickers 448[9] which has also found wide acceptance in British gas-turbines during the past few years.

Steels of this general class demand precise control in composition, quality, and manufacturing procedure, if reproducible performance is to be achieved. They have a strong tendency to retain the ferritic "δ-phase", which when present to excess, can produce a serious lack of ductility at low temperatures in the transverse direction of forging. This is particularly troublesome in large forgings. Naturally, the engineer must fix minimum requirement figures for percentage reduction-of-area and percentage elongation in test-pieces cut from disk forgings, but it has been found that a dispersion of retained "δ-phase" makes the maintenance of minimum requirements a difficulty. Apart from the control of the alloying constituents, the gas contents are now measured; nitrogen being added to minimize "δ-phase" formation.

These newer British steels (H.46 and F.V.448) possess the desirable characteristics of a low coefficient of thermal expansion, while retaining a high ratio of 0·1% proof stress to the maximum-stress. They also have completely adequate scale-resistance, outstanding creep-strength, freedom from rusting on standing, high strength–weight ratio, and freedom from undue growth on cold-spinning when used for turbine disks.

Later, a steel was requested to meet the very exacting specification of less than 0·1% creep strain in 100 hours at 500°C at a stress of 32 tons/in^2 (50 kg/mm^2). Two steels have been developed which meet this specification, namely, Jessop H.53 and Firth-Vickers 535. These steels are oil-hardened and tempered and are usually used at a tensile strength level of 70–75 tons/in^2 (110–119 kg/mm^2). They are specially suitable for applications where very high working stresses are involved in the temperature range 400°–575°C. Additions of cobalt were found to be effective; but, again, the composition must be carefully balanced to retain adequate ductility and freedom from embrittlement. The metallurgy of steels of this class is highly complex, and many features of them are not yet fully understood. Nevertheless, they have played, and are continuing to play, a vital rôle in advanced power-units for aircraft. In some recent turbo-jets of high power rating, it is becoming increasingly difficult to avoid disk rim-temperatures rising to above 575°C, which represents a good maximum temperature limit for the above steels. It may therefore become necessary to turn again to austenitic steels, in spite of their higher cost and added weight. One of these, "Discalloy", a precipitation-hardening steel, has been widely used with success in the United States. Others include A.286 and G.56 (Table 1).

3. ALLOYS FOR FLAME-TUBES

When the aircraft gas-turbine was first invented the requirements of materials for flame-tubes were very ill-defined, but it was apparent that high temperatures would be reached and that therefore good oxidation resistance would be necessary. Austenitic stainless steels of the 18 : 8 Cr/Ni type were first used, and although reasonable performance resulted, failures were soon experienced by buckling and cracking at air-holes and welds.

At this time a nickel–chromium alloy (Henry Wiggin's Nimonic 75) became available, and in view of its known high oxidation-resistance this material was tried. Vastly improved results were obtained, and this alloy has since remained the standard material for combustion systems. The nickel–iron–chromium alloy, "Inconel", was also tried in the early days, but its performance was classed as intermediate between those of the austenitic steels and Nimonic 75.

The marked increase in life of these components in modern engines has largely been achieved by improvements in design, particularly regarding the correct use of cooling air. Nimonic 75 has proved an admirable material on account of its ready fabrication and ease of weldability. Failures of flame-tubes are generally traced to one of two causes; either distortion, or cracking due to thermal-fatigue. The distortion arises from stresses caused by temperature-gradients, and thermal-fatigue is caused by rapid changes of temperature producing steep temperature-gradients. The thermal-fatigue or thermal-shock properties of sheet metals have been measured by Lardge[10], and his results showed that Nimonic 75 had comparable thermal-fatigue properties to the stainless steels (see Table 2).

TABLE 2

Thermal fatigue tests on sheet alloys

(after H. E. Lardge)

Alloy	Sheet thickness (inch)	Average number of cycles to crack (Temperature range 900° to 50°C)
18/8	0·048	380
DTD 493	0·048	300 (extrapolated)
Inconel	0·051	215
Nimonic 75	0·048	220
Nimonic 80A	0·049	232
Nimonic 90	0·052	140

The relative merits of different materials vary markedly with the testing conditions, and no satisfactory test directly related to the service conditions has yet been devised. The test-results are therefore to be regarded as only an approximate indication of merit, and the final assessment of a particular material must be made under service conditions. Austenitic steels of the 25 : 20 Cr/Ni variety (e.g. Red Fox 31) have also been successfully used for flame-tube construction, but again their "life" is intermediate between that for 18 : 8 austenitic steels and Nimonic 75.

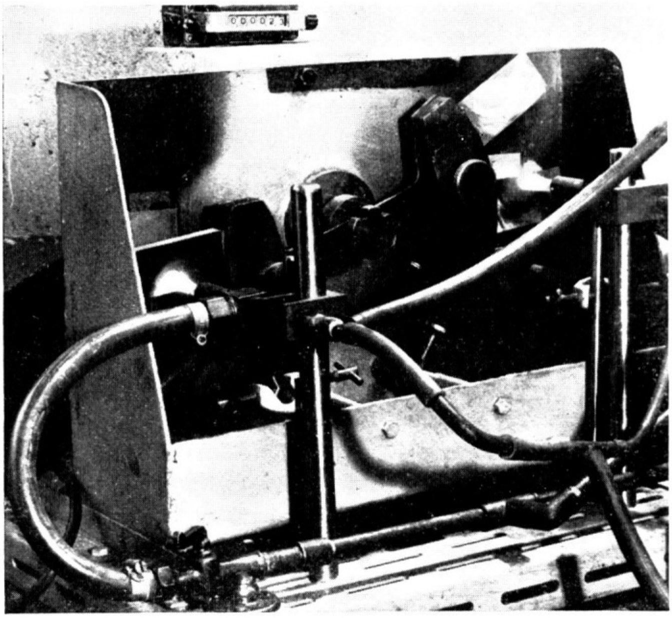

FIG. 2. Thermal-shock or thermal-fatigue testing rig.

Figure 2 shows one rig for carrying out "thermal-shock" or "thermal-fatigue" tests on a heat-resisting material. Four wedge-shaped test-pieces on a time-controlled rocking-arm alternately present two test-pieces to the pressurized burners. Rapid heating and cooling can be arranged with attendant oxidizing conditions, and the number of cycles of heating and cooling required to produce a given (arbitrary) degree of edge-cracking can be observed, and is taken as an index of the thermal-shock resistance. Comparative tests are made between a new alloy, and one, say, of proven value in service. In one laboratory, the test consisted of heating a wedge-shaped specimen having a trailing-edge thickness of 0·020 in. (0·50 mm) in a town's gas-flame for a period of one minute, and then allowing it to cool freely in air for 1 min. This is considered to be a complete cycle.

The "thermal-shock index" is known to be affected by the thickness of the trailing-edge, the maximum temperature attained, and the rates of heating and cooling. In one group of tests on eight materials for a tempera-sure range of 950°C to 450°C, the index varied between 114 and 1045, so that appreciable differences can be observed.

If future designs of flame-tubes result in higher operating temperatures it is likely that sheet materials with increased high-temperature strength will be required. Some experiments have been carried out on the use of Nimonic 80 and Nimonic 90 sheet, but the greater difficulty of fabrication when using these materials, makes their adoption uncertain. Furthermore, with Nimonic 75, welded-joints have strengths equal to that of the parent material; but the precipitation-hardening alloys, Nimonic 80 and Nimonic 90, normally lose considerable strength in the weld region. Although this can be recovered by the use of suitable electrodes, and by heat-treatment after welding, the practical difficulties involved are considerable. Compositions of the materials referred to are given in Tables 3 and 5.

TABLE 3

Nominal compositions of sheet alloys for flame tubes (weight per cent)

Alloy	Element				
	C	Cr	Ni	Ti	Fe
18 : 8	0·12	18	9	0·5	Balance
DTD 493	0·20	25	15	—	Balance
Inconel	0·1	14	80	—	Balance
Nimonic 75	0·1	20	Balance	0·3	2·5

TABLE 4

Nominal compositions of alloys for inlet-nozzle guide-vanes (weight per cent)

Alloy	Element								
	C	Cr	Ni	Co	Fe	W	Mo	Nb	Ta
H. R. Crown Max. (F.V.)	0·25	23	12	—	Bal.	3	—	—	—
Vitallium	0·2	30	—	65	—	—	5	—	—
G. 39 (J.)	0·52	20	Bal.	—	—	3	3	1·5	1·5
C.242 (R.R.)	0·3	20	Bal.	10	—	—	10	—	—

TABLE 5

Nominal compositions of Nimonic alloys for turbine blades. (Henry Wiggin & Co. Ltd.) (weight per cent)

Alloy	Element						
	C	Cr	Co	Mo	Ti	Al	Ni
Nimonic 80A	0·05	20	—	—	2·4	1·2	Bal.
Nimonic 90	0·08	20	17	—	2·4	1·2	Bal.
Nimonic 95	0·08	20	17	—	3·0	1·8	Bal.
Nimonic 100	0·20	12	20	5	1·5	4·5	Bal.

4. INLET-NOZZLE GUIDE-VANES

These components of a gas-turbine are subjected to the highest operating temperatures in the engine, since they are in the direct path of the combusted fuel with its excess air. The primary requirement of the constructional material is therefore oxidation and corrosion-resistance, with sufficient inherent thermal-fatigue resistance to withstand rapidly changing temperatures. The mechanical strength demands are relatively small since only gas-reaction stresses are involved.

It has been suggested that a suitable criterion is that creep deformation should not exceed 0·5% under a stress of 2·0 tons/in^2 (3 kg/mm^2) at 1000°C. Both cast, wrought, and fabricated components have been used, and many different alloys have proved reasonably satisfactory. Alloyed austenitic steels such as Firth-Vickers H.R. Crown Max were used for cast inlet-nozzle guide vanes in the early days, primarily on account of their good casting qualities. Cobalt-base alloys (e.g. Vitallium) have also been used, but high-temperature-strength considerations have led to complex nickel-base alloys now being generally preferred. The Nimonic alloys, including 75, 80 and 90, have all been used from time to time in the wrought condition, frequently being supplied as extruded and cold-rolled sections, to minimize the necessary machining. Overall cost considerations, however, have generally led to the adoption of castings. Although Nimonic 80 and 90 have been extensively used in the cast form; the presence of titanium and aluminium makes refined production techniques essential. Nickel–chromium-base alloys such as Jessop G.39 and Rolls-Royce C.242, which are free from titanium and aluminium, are more readily handled in the foundry; and are consequently favoured for this type of component in which the highest elevated temperature strength is not required. Typical compositions are given in Table 4.

Increased gas-inlet temperatures, for higher thermodynamic efficiency in the turbine, are leading to the use of cooled nozzle-blades, and these are manufactured from alloys of the type referred to above; either by casting, by extrusion, by fabrication from sheet, by powder metallurgy, or by other means. These methods are referred to in greater detail in a later section.

5. TURBINE-BLADE MATERIALS

Since about 1944, the turbine-blades of all aircraft gas-turbines in Great Britain have been made in wrought precipitation-hardening nickel-chromium alloys of the Nimonic series. The first alloy of this type, Nimonic 80, consisted of the oxidation-resistant alloy 80 : 20 Ni/Cr with additions of $2\frac{1}{2}\%$ Ti and 1% Al. This alloy had short-time creep, and stress-rupture, properties appreciably higher than those of the special austenitic steels then available.

Progressive developments of the Nimonic(23, 24) alloys have resulted in steady improvements in their creep-strengths. The changes in composition have included modifications of the matrix by the incorporation of cobalt and molybdenum, and by changes in the contents of titanium and aluminium, which give rise to the precipitation-hardening. Four principal alloys in the series have been used for turbine blades, and these are detailed in Table 5. The stress-rupture properties of these alloys in terms of the 100 hr rupture-stress at various temperatures are given in Table 6. It is seen from this table that progressive improvement of the high-temperature strength has been obtained, resulting in an increase of the permissible operating temperature for aircraft gas-turbines of approximately 120°C. In addition to the major modifications of composition outlined in Table 5 progress has also been made by a careful study of the methods of heat-treatment(25), and by developments in the methods of melting, and in the hot-working of the alloys. Many design data(26) are now available.

TABLE 6

Stress–rupture properties of Nimonic alloys
(Henry Wiggin & Co. Ltd.)

Alloy	Stress-to-rupture in 100 hr (kg/mm)			
	(1380°F) 750°C	(1500°F) 815°C	(1600°F) 870°C	(1725°F) 940°C
Nimonic 80A	28	16	—	—
Nimonic 90	34	20	11	—
Nimonic 95	36	23	15	—
Nimonic 100	40	28	20	11

Table 6—Continued page 980.

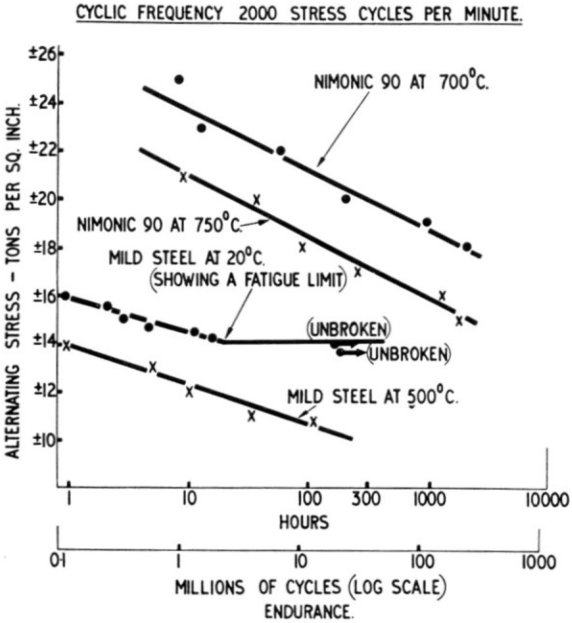

FIG. 3. Hot-fatigue endurance results, extracted from miscellaneous researches at the Mechanical Engineering Research Laboratory, East Kilbride, Glasgow.

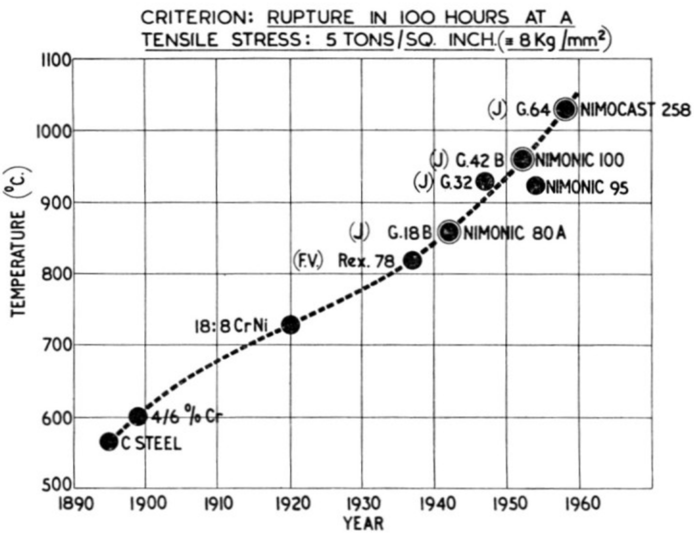

FIG. 4. Curve showing progress in creep-resisting alloys on a common basis.

TABLE 6—*continued*.

Alloy	Stress-to-rupture in 100 hr (tons/in^2)			
	(1380°F) 750°C	(1500°F) 815°C	(1600°F) 870°C	(1725°F) 940°C
Nimonic 80A	18	10	—	—
Nimonic 90	22	13	7	—
Nimonic 95	23	14	10	—
Nimonic 100	25	18	13	7

Besides the stress-rupture properties at elevated temperatures of which some general idea of the progress achieved can be seen in Fig. 4, the "fatigue" characteristics are also of considerable importance in connexion with turbine blades. It has been shown by Allen and Forrest(11) that the high-temperature fatigue properties under certain simple conditions of stressing are in direct relationship with the stress-rupture characteristics. Although it would be unwise to apply this generalization to all conditions of fatigue stressing; it may well prove a useful guide particularly in development work. As regards "hot-fatigue" the point should be made that while many materials at ordinary ambient temperatures show a "fatigue limit", almost independent of the frequency of stressing; at elevated temperatures it is only possible to quote an "endurance" figure. This, in turn, is closely linked with the "time-of-testing" as well as the total number of the applied cycles of stress required to cause failure. Furthermore, the "frequency-of-stressing" starts to become a test-variable, as at the higher frequencies-of-stressing rather more optimistic results are obtained(12). Ignoring this last factor for the moment, the above matters are well illustrated by Fig. 3 which summarizes recent data obtained by the Mechanical Engineering Research Laboratory, East Kilbride, Glasgow. In quoting "endurance limits" for a given range of alternating stress it is desirable to specify as well the time-of-test (in hours) to failure. At high temperatures both simple materials like mild-steel and complex alloys like Nimonic 90, show steadily falling endurance curves when the alternating-stress range is plotted against the logarithm of either "time-to-failure" or the aggregate number of "cycles-to-failure".

During recent years, much effort has been expended in trying to elucidate the effects of complex stressing, particularly under creep-conditions. Johnson *et al.*(13,14) have shown that for a 0·5% molybdenum steel at 550°C, and for commercial copper at 250°C; their main deductions were that while the Mises criterion of plastic flow characterizes "primary creep", a most important part in the development of "tertiary creep" towards eventual failure is played by the *maximum tensile principal stress*, presumably through the medium of the propagation of cracks, incipiently created by primary creep. Evidence is now being gained that materials which do not crack in the tertiary range of creep, follow the Mises–Hencky criterion up to fracture.

6. RECENT ADVANCES AND NEWER METHODS (TURBINE BLADES)

For many years now turbine blades have been produced from forgeable alloys by the orthodox methods of producing pre-forms followed by drop-stamping, or final forging in precision dies in large crank presses. The methods used have either invoked final machining and grinding for an over-size forging, or the forging operation itself has been carried out sufficiently closely to enable the blade-form to be put into use after minor rectification by grinding or polishing.

During the past five years, efforts have been made to reduce the cost of blade production as well as to increase the blade operating temperature. Thus in some turbines, precision-cast blades by the lost-wax process have ousted forgings for a short period; only for forgings to oust precision-castings when a better alloy became available. Nevertheless, forgings have held the field in the vast majority of instances, but more recently engineers

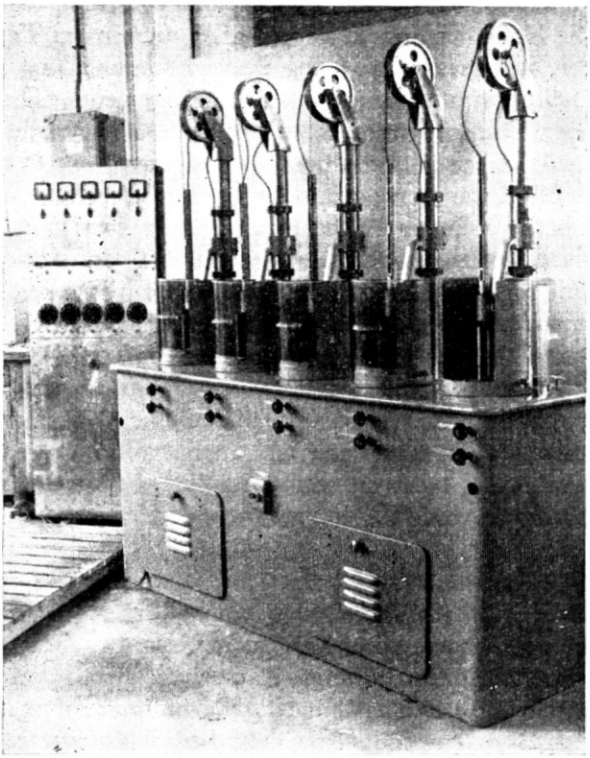

FIG. 5. General view of "Penatron" deep-hole spark-drilling equipment. (By courtesy of Messrs. Burton, Griffiths & Co., Ltd.)

have favoured cooling methods to cope with the increase in turbine inlet gas temperatures. In consequence, a need has arisen for hot-worked blades of precision contour having fine longitudinal holes running through the body of the blade structure for carrying cooling air, usually bled from the compressor system. One method of achieving this has involved the drilling of a pre-form with several quite large holes followed by the plugging of these holes with some metal or material readily dissolved in acid, and followed by hot extrusion of the combination, so that continuous ducts are ultimately produced inside the blade itself. Much skill and ingenuity have been expended on these products, and some have utilized in extrusion the liquid-glass method of lubrication associated with the Ugine-Sejournet process. In this process it has been found difficult to ensure ducts of exactly the same cross-section, such is often desirable. Cast blades have also been made with numerous small passages produced by the use of silica-tube cores, which are subsequently dissolved by molten alkalis. Other methods have emerged after further study, and one which is showing promise involves the deep-hole drilling of fine holes (e.g. 0·060 in. diameter) by a new spark-machining method[15] evolved in the author's laboratory. One equipment for this purpose has been called the "Penatron" and is seen in Fig. 5. This machine is capable of drilling holes 0·060 in. diameter in the Nimonic alloys at a speed of about 6 in/hr. The method is gaining favour, and means are being devised for polishing the internal bore of the holes by miniature grinding arrangements. The method is only applicable where straight holes from one end of the blade to the other can be arranged, and usually the operation is carried out from the tip of the blade to a blind transverse channel located in the blade root.

In cases where cooling holes have to follow a curved contour, other methods are clearly necessary. One method is to form the holes in a simplified blade form, to fill them with, say, loose iron wires, and then to forge the combination to the final more complex shape. The filler wires are then extracted by acid treatment, and the additional complication of curved cooling passages has been achieved. By these means, inlet gas temperatures can be tolerated about 200°C higher than is possible with an uncooled blade of the same material.

Much attention has been devoted recently by three British firms to powder-metallurgy techniques in order to develop a good method of making blades.

The Powder-metallurgy Approach

For over ten years powder metallurgists in Great Britain have envisaged the possibility of fabricating complex blade shapes from powders, having "built-in" cooling ducts, and needing only the minimum of final machining. At first, the target was to produce a sintered blade equivalent to a forged blade, and to compare the major properties at both low and elevated temperatures. The early work was most disappointing; as clean, fine,

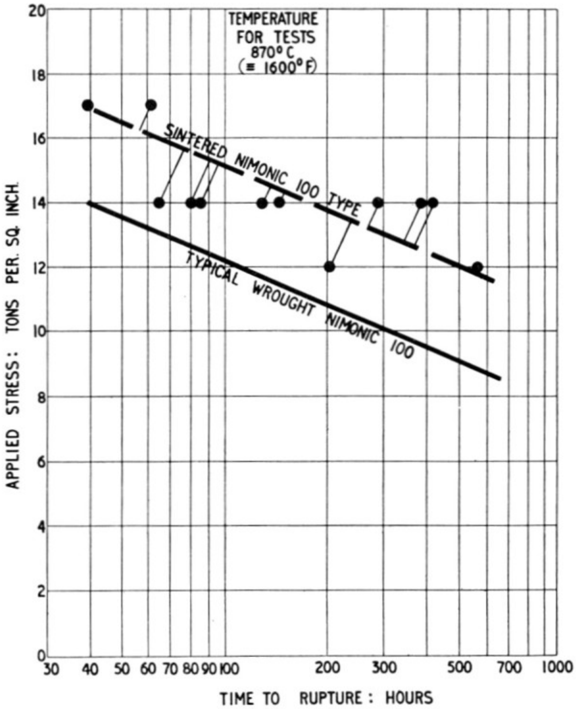

FIG. 6. Stress/time-to-rupture data comparing wrought Nimonic 100 and similar sintered materials (B.S.A. data).

pre-alloyed powders were practically unobtainable, and those which were, gave most disappointing mechanical properties after sintering in dry hydrogen. Further progress was then made by melting brittle binary and ternary alloys, either in air or *in vacuo*, followed by mechanical attrition and prolonged grinding in rod and ball mills until fine powders were available after sieving. Combinations of these in suitable blends to ensure the correct composition gave much improved results, but the properties were still inadequate for blade manufacture.

During the past three years Watkinson of the author's Company has succeeded in producing high quality pre-alloyed powders[16] by atomization in atmospheres which gave the minimum of oxidation and contamination. These powders when pressed in dies between 30 and 50 tons/in^2 and subsequently sintered *in vacuo* for periods ranging between 2 and 20 hr at 1300°C have yielded promising creep and hot-fatigue properties. For example, taking the (comparatively new) wrought alloy (Nimonic 100) and comparing it, at 870°C, with a sintered component built-up from fine pre-alloyed atomized powder; has furnished the following ratio properties. (Table 7.) Sintering was for 20 hr *in vacuo* at 1300°C.

TABLE 7

Ratio properties

Specimen Test temperature: 870°C	Stress-to-rupture	Hot-fatigue: endurance
Typical wrought "Nimonic 100"	1·0	1·0
Pressed and sintered "Nimonic 100 Type" powder	1·3	0·75–0·90 (according to process)

Some stress vs. time-to-rupture data are shown plotted in Fig. 6. It is of interest to note that the density of the sintered component varies with the powder batch and is usually a few per cent lower than the equivalent wrought alloy, which shows that the sintering process has proceeded to a very advanced stage. Subsequent hot-forging did not readily increase the density. The wrought and sintered alloys had a very similar micro-structure except that the grain size of the sintered material was much finer and more uniform. One comparison of grain sizes gave an *average* grain "diameter" of about 0·14 mm for the wrought material as against 0·03 mm for the pressed and sintered material. These figures correspond on the A.T.S.M. grain-size chart to ratings of 2–3 for the wrought, and

Typical wrought "Nimonic 100" Typical sintered "Nimonic 100 type"

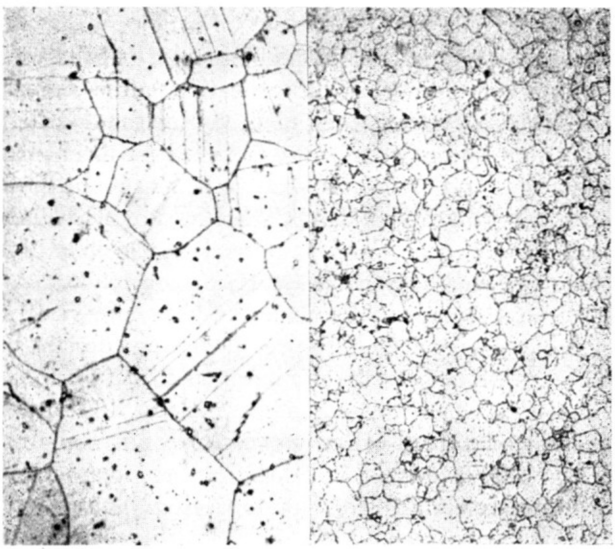

FIG. 7. Showing grain-size comparisons (×100) between wrought and sintered materials (B.S.A. data).

8 for the sintered specimens. The photomicrographs are reproduced in Fig. 7. The sintered component has slightly reduced mechanical toughness as measured by impact tests at room temperature, but usually physical properties are obtained which are acceptable to the engineer. The conclusion can now be drawn that by using pre-alloyed atomized powders of high quality, with pressing and sintering procedures which are not unduly difficult; it is now possible to present the engineer with a turbine blade having a balance of properties which are likely to be satisfactory and which will meet turbine service conditions. Moreover, the way is also opened up for the introduction of cooling ducts and channels in alloy compositions which are known to be almost unforgeable, but which nevertheless can be produced via the powder metallurgical approach. Furthermore, "uses" can be made from powders, and when the component is adequately protected, a measure of forging or hot-coining can be imposed under a crank-press. In turn, this reduces demands on machining facilities with subsequent rectification, and ensures fully forged structures of small and controlled grain-sizes. Once these features are appreciated, many developments on these lines can be expected in the immediate future.

Turning now to the more formal aspects of the subject, reference must be made to the early work of Buswell *et al.*[17] in the production of Vitallium blades and test-specimens from powder, mainly by mixing the elemental constituents. This early work has been followed up by producing blades in various alloys and having cooling passages introduced by their elegant cadmium-wire technique, whereby the cadmium wires are introduced into the "green" pressing[18]. Prior to final sintering, or even pre-sintering, the cadmium is removed by volatilization *in vacuo* at about 650°C. Thus one successful method for introducing ducts of controlled cross-section and length in a powder-metal blade has been devised and proved in practice on a limited scale. Modifications of the method have followed, but cadmium has still played a notable part even in the modified methods. It will also be apparent that sintered blade-blanks can be drilled by spark-machining, either at an intermediate stage in the process, or in final fabrication. These methods, taken alone or in combination, have opened up good prospects of making highly complex structures relatively easily and economically, while satisfying many basic mechanical requirements in relation to creep and fatigue at elevated temperatures.

Precision Casting and Vacuum Techniques

The view has been held for some time that if the quality of precision lost-wax castings could be raised they would prove more competitive with precision forgings, which in turn have become more and more difficult to produce as alloys have become stronger and more "sensitive" to hot-working.

Precision castings have been adopted more boldly in the United States than in Great Britain and, even when air-melting methods only are taken into account, the progress made has been impressive.

Credit should be paid to the United States for pioneering the vacuum melting and casting of the more complex heat-resisting alloys on a tonnage basis. In particular, the work of Darmara at Utica has made metallurgical history.

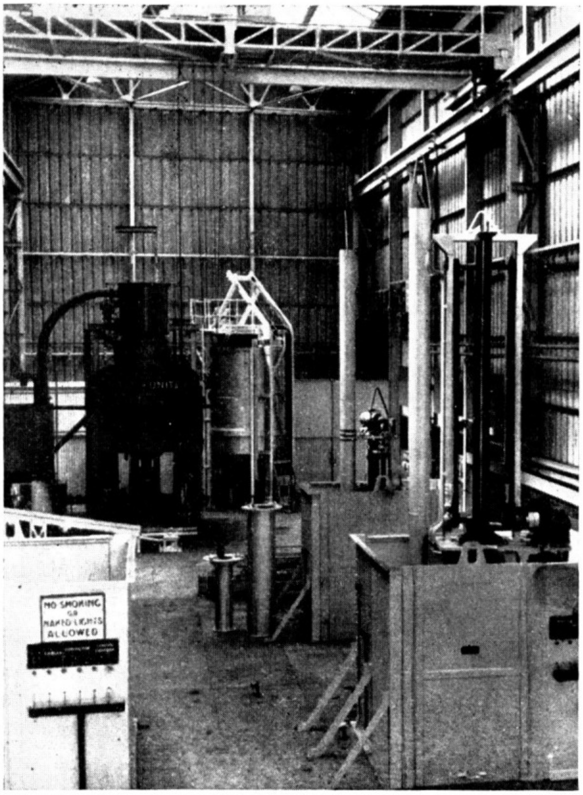

FIG. 8. View of modern consumable-electrode vacuum arc-melting plant. (By courtesy of Messrs. Wm. Jessop & Sons, Ltd.)

In Great Britain, since 1952, I.C.I. Ltd. and Wm. Jessop & Sons Ltd. have installed consumable-electrode vacuum arc-melting equipment (Fig. 8). While the Jessop plant was installed primarily for titanium alloy melting, it has subsequently been adapted to the vacuum-melting and casting of ingots in other materials, including steels and high-temperature alloys. Forgings and bar stock of high quality have now become available having reduced gas contents, greater freedom from non-metallic inclusions; and enhanced ductility, creep strength and endurance at elevated temperatures. Vacuum-melting and casting are now being applied to lost-wax precision castings in the newer alloys which demand the refinement of freedom from oxidation. Refractory mould problems arise when used *in* *vacuo*, but these difficulties are also being surmounted. In consequence, vacuum-melted precision castings are now being considered for the more advanced turbines under development, especially where blade-cooling is being side-stepped. Both "Nimocast 258" and "Jessop G.64" are under test, the former being air-melted and the latter being vacuum-melted and cast (Fig. 4). A wide range of nickel-base casting alloys[19] is now available. A tribute must be paid to the pioneer work of Rolls-Royce who as early as 1948 had evolved nickel-base alloys containing molybdenum (e.g. C.88 and C.242) and who thus pointed the way for much subsequent progress. For example, the major overhaul life in civil use of the cast nozzle-guide-vanes used in the Rolls-Royce "Dart" has risen from an initial period of 450 hr to the present 2000 hr. This is equivalent to about 700,000 miles of flying between overhauls, or about 6 months' working. The consequent economic advantages of high aircraft utilization are apparent.

7. THE FUTURE

Rocket material requirements have focused attention on high-melting-point metals such as molybdenum, tungsten, tantalum and niobium. So far the higher-temperature shorter-life requirements of the rocket have not noticeably benefited the aero gas-turbine, but the molybdenum alloys are possibly significant to both fields. The problems and achievements involved in protecting molybdenum-rich alloys against serious oxidation by the trioxide MoO_3 have been ably summarized by Harwood and Semchyshen[20] who see great promise in molybdenum-base alloys for high temperature service at over 1000°C, particularly after more intensive development.

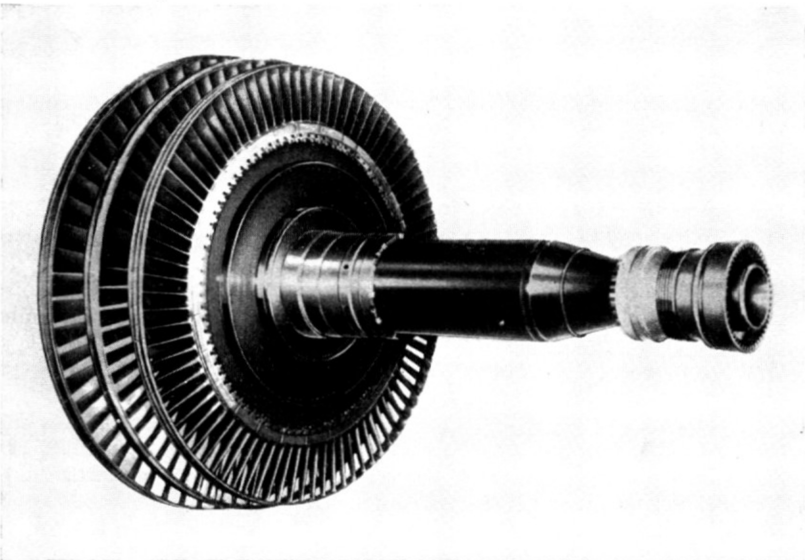

FIG. 9. General view of modern three-stage turbine assembly. (Reproduced by kind permission of Messrs. Rolls-Royce Ltd.).

It is still fundamentally true that the engineer prefers a material with the necessary inherent properties, but so far the choice has been in favour of tried and tested materials such as the nickel–chromium–cobalt-base materials having small passages for cooling air in addition. The next step might be porous structures[21], adequately cooled, but made in alloy powders of greater strength than were available in 1954. With suitably reinforced porous components the great advantages of transpiration cooling, or even sweat cooling, might then be exploited. It will be some time before this stage is reached, as the air-cooled turbine era is only just beginning[22]. Newer alloys will certainly arise, but economic considerations will winnow the alternatives. The advent of vacuum-melting and casting will up-grade the performances and reliabilities of known materials, and will enable higher percentages of some constituents to be added, thereby greatly enhancing the properties. The future, therefore, is full of latent promise and, as ever, will greatly reward those who storm its strongholds.

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