High Temperature Structural Materials for Gas Turbines *

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Abstract
A general overview of hot components materials developed for both aeroengines and industrial gas turbines is firstly given; these materials are at present near exclusively Ni-and Co-base alloys.
A brief illustration of margins for further developments of metallic materials, and of ceramic and composite materials potentials is completed by an evaluation of the most relevant work in progress and problems arised.

Riassunto
Vengono esaminati i materiali sviluppati per i componenti a caldo sia di turbine a gas per impieghi aeronautici che per impieghi industriali che sono quasi esclusivamente leghe a base di Ni e di Co.
Dopo una breve illustrazione dei margini per ulteriori sviluppi dei materiali metallici e delle potenzialità dei materiali ceramici e dei compositi vengono valutati i lavori in corso più rilevanti ed i problemi incontrati.

Gas turbines

Gas turbine machines are thermodinamically based on the Brayton cycle, Fig. 1. The cycle processes take place in the three main machine components: compressor, combustor and turbine. The expansion of the working fluid may occur entirely within the turbine, which is the case of the industrial gas turbine, or partially in the turbine and the remaining in the jet nozzle, which is the case of the aero-engine.

The first successful design of an industrial gas turbine is attributed to Aegidius Elling (1923) whereas the first aeroengine to power an aircraft (He 178, 1939) has been designed by Hans Joachim Pabst. Since then the gas turbine technology is characterized by a history of tremendous advances in power output and efficiency, where a key role has been played by the turbine inlet temperature as it is shown in Fig. 2. In the last 40 years military engines have increased the thrust-to-weight ratio to about 4 times the initial value, the commercial engines have decreased their specific fuel consumption of about 60%, the industrial gas turbines have rised their efficiency (simple cycle) from 20% up to about 35%. The advancements have been possible by a continuous increase in turbine entry temperature with a rate of about 15 K per year: Fig. 3 illustrates the turbine temperature climb. Among the most advanced machines under development, two projects involving Fiat Avio as one of the responsible partners are: the commercial aeroengine GE 90 and the industrial gas turbine FMW 701F.

Turbine temperature is strictily related to material capability and this implies a close connection between materials evolution and gas turbine technology progressive growth. Fig. 4 shows the GE 90 configuration and Fig. 5 shows the FMW 701F turbine section. The aim of the present paper is to illustrate the state-of-the-art in the field of gas turbine high temperature structural materials as well as the lines of future developments.

Requirements

Combustion chamber

The majority of combustion hardware is currently fabricated from superalloy sheets. The overriding material properties are formability and weldability for initial manufacture as well as for ease of repairing service-induced defects. The metal temperature is kept under the allowable limit by diluting the combustion products, cooling the flame tube walls and applying thermal barrier coating. The selected materials must be resistant to thermal fatigue, oxidation and carburization, and be metallurgically stable in service; moderate creep and proof strengths are sufficient to avoid buckling.

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Nozzle guide vanes

Nozzle guide vanes direct the hot gases into the rotating stage of the turbine at the most favorable angle of incidence. Specially for the first stage vanes the hot gas impingement makes necessary a high degree of cooling in order to obtain an adequate service life. The complexity of the cooling configuration, shown in Fig. 6a, requires casting technology to fabricate the component. The aerodynamic shape requirement does not allow the same degree of cooling in all the component areas, so that uneven heating and cooling are cause of thermal fatigue cycling in particular at both leading and trailing edges. Moreover, a good creep strength material is needed since in steady state operation the leading edge experiences the highest temperature as well as the highest gas bending stress. In conclusion the major mechanical requirement is thermal fatigue resistance, followed by creep strength. Besides an adequate oxidation/hot corrosion resistance is also necessary: this property is obtained by the intrinsic material resistance enhanced by the application of protective coatings. The corrosion problem is more severe in the industrial gas turbine since very long service lives are required as well as the option to burn a wide range of fuels, from natural gas to residual and crude oil and coal derivatives. The component repairability is generally a requirement.

Blades

The rotating turbine blades have the task to extract energy from the hot gases transforming it in mechanical energy able to drive the compressor and, in the case of industrial gas turbines, also the generator. In addition to the severe conditions mentioned for nozzle guide vanes, blades are strongly centrifugally loaded and are more subject to vibratory phenomena. As a consequence all the adverse phenomena are to be considered as major items: creep, thermal fatigue, high cycle fatigue and oxidation/ hot corrosion. Cast parts are generally used, at least for high pressure turbine blades, since they present superior mechanical properties and can be fabricated more easily into complex, highly cooled configurations (Fig. 6b).

Materials in use

Nickel and Co-base superalloys constitute about 40% of a modern aeroengine total weight: in particular combustor and turbines are entirely made in superalloys. A similar situation does exist for the aeroderivative industrial gas turbines, whereas superalloys are limited to the hot gas exposed parts in heavy duty gas turbines.

Table 1 gives a summary of the alloys, and related nominal compositions, used for combustion chambers, exhaust units and reheat liners. It is evident how the number of alloys is rather limited: the same alloys are used in both aeroengines and industrial gas turbines.

TABLE 1 - Combustion chamber alloys and their composition

ALLOY	Ni	Cr	Co	Mo	W	Al	Ti	Fe	C	
Hastelloy X	Bal.	22.	1.5	9.0	0.6			18.5	0.07	Si 0.4 Mn 0.6
Nimonic C263	Bal.	20.	20.0	5.9		0.45	2.15	0.7	0.06	Si 0.25 Mn 0.4
Nimonic PK33	Bal.	18.	14.0	7.0		2.1	2.2		0.05	
Inconel 617	Bal.	22.	12.5	9.0		1.0			0.07	
Haynes 188	22.0	22.	Bal.		14.0				0.1	La 0.05
Nimonic 86	Bal.	25.		10.0					0.05	Mg0.015 Ce0.03

Table 2 gives a summary of the most important alloys used in industrial gas turbine nozzle guide vanes, all of them are cast alloys.

TABLE 2 - Nozzle guide vane alloys and their composition

ALLOY	Ni	Cr	Co	W	Ta	Nb	Al	Ti	Fe	C	B	Zr
X 40	20.0	25.0	Bal.	8.0		1.0			1.0	0.5	0.01	
X 45	10.0	25.0	Bal.	8.0					1.0	0.25	0.01	
MAR M 509	10.0	23.5	Bal.	7.0	3.5			0.2	1.0	0.6	0.01	0.5
FSX 414	10.0	29.0	Bal.	7.5					1.0	0.25	0.01	
ECY 768	10.0	23.5	Bal.	7.0	3.5		0.15	0.2	1.0	0.60	0.01	0.05
IN 939	Bal.	22.4	19.0	2.0	1.4	1.0	1.9	3.8		0.15	0.01	0.1
GTD 222	Bal.	22.5	19.0	2.0	1.0	0.8	1.20	2.3		0.1	0.01	

Table 3 summarizes the alloys developed to manufacture turbine blades by forging. The last alloys have been developed in the early 1970s, their application in advanced machines is limited to the industrial gas turbines end stages.

TABLE 3 - Blade wrought alloys and their composition

ALLOY	Ni	Cr	Co	Mo	W	Nb	Al	Ti	Fe	C	B	Zr
INCONEL-X750	Bal.	15.5				1.0	0.7	2.5	7.0	0.04		
NIMONIC 105	Bal.	15.0	20.0	5.0			1.2	4.7		0.13	0.006	0.1
UDIMET 500	Bal.	18.0	18.5	4.0			2.9	2.9		0.08	0.006	0.05
INCONEL 700	Bal.	15.0	17.0	5.0			4.0	3.5		0.06	0.03	
NIMONIC 115	Bal.	14.5	13.0	3.0			5.0	3.8		0.15	0.016	0.04
UDIMET 520	Bal.	19.0	12.0	6.0	1.0		2.0	3.0		0.05	0.005	
UDIMET 710	Bal.	18.0	15.0	3.0	1.5		2.5	5.0		0.07	0.02	
UDIMET 720	Bal.	17.9	14.7	3.0	1.3		2.5	5.0		0.03	0.033	0.03

Table 4 gives a summary of the main cast alloys developed for turbine blades and largely used in the last 30 years: the alloys with high Cr content (> 14%) are preferred for industrial gas turbines, whereas alloys with high content of Al, Ti and Nb as well as Mo, W and Ta are generally preferred for aeroengines.

The last two decades have been characterized by the introduction of the directional solidification (DS) casting process to produce blades and vanes with columnar grains aligned parallel to the principal-stress axis. The DS superalloys which have been successfully and extensively adopted by aeroengine companies are René 80H, Mar-M-247 and Mar-M-002. GTD-111 is the first DS alloy introduced in an industrial gas turbine.

TABLE 4 - Blade/nozzle guide vane cast alloys and their composition

ALLOY	Ni	Cr	Co	Mo	W	Ta	Al	Ti	C	B	Zr	
UDIMET 500	Bal.	18.0	19.0	4.2			3.0	3.0	0.07	.007	0.05	
INCONEL 713 LC	Bal.	12.0		4.5			5.9	0.6	0.05	0.01	0.1	Nb 2.0
RENÈ 77	Bal.	14.6	15.0	4.2			4.3	3.3	0.07	.016	0.04	
IN 100	Bal.	10.0	15.0	3.0			5.5	4.7	0.18	.014	0.06	V 1.0
RENÈ 80 (80H)	Bal.	14.0	9.5	4.0	4.0		3.0	5.0	0.17	.015	0.03	(Hf0.8)
MAR-M-246	Bal.	9.0	10.0	2.5	10.0	1.5	5.5	1.5	0.14	.015	0.05	
IN 738 LC	Bal.	16.0	8.5	1.70	2.6	1.70	3.4	3.4	0.10	0.01	0.05	Nb 0.9
MAR-M-247	Bal.	8.3	10.0	0.7	10.0	3.0	5.5	1.0	0.14	.015	0.05	Hf 1.5
MAR-M-002	Bal.	9.0	10.0		10.0	2.5	5.5	1.5	0.14	.015	0.05	Hf 1.5
RENÈ 125	Bal.	8.5	10.0	2.0	8.0	3.8	4.8	2.5	0.11	.015	0.05	
GTD 111	Bal.	14.0	9.5	1.5	3.8	2.8	3.0	4.9	0.10	0.01		

A further development of the DS process is the single crystal (SC) casting technology, nowadays widely used in aeroengine blades. Table 5 summarizes the known SC alloys. The last four listed alloys are worth of a particular mention since PWA 1484, CMSX-4 and René N5 are the so called "second generation alloys", developed respectively by Pratt & Whitney, Cannon-Muskegon and General Electric, while SC16 is being developed by ONERA for applications in industrial gas turbines.

TABLE 5 - Blade single crystal alloys and their composition

ALLOY	Ni	Cr	Co	Mo	W	Ta	Nb	Al	Ti	Hf	Re	V
PWA 1480	Bal.	10.0	5.0		4.0	12.0		5.0	1.5			
RENÈ N4	Bal.	9.0	8.0	2.0	6.0	4.0	0.5	3.7	4.2			
SRR 99	Bal.	8.0	5.0		10.0	3.0	0.7	5.5	2.2			
RR 2000	Bal.	10.0	15.0	3.0				5.5	4.0			1.0
AM1	Bal.	7.0	8.0	2.0	5.0	8.0	1.0	5.0	1.8			
CMSX-2 (-3)	Bal.	8.0	5.0	0.6	8.0	6.0		5.6	1.0	(0.1)		
CMSX-6	Bal.	10.0	5.0	3.0		2.0		4.8	4.7	0.1		
PWA 1484	Bal.	5.0	10.0	2.0	6.0	8.7		5.6		0.1	3.0	
CMSX-4	Bal.	6.4	9.6	0.6	6.4	6.5		5.6	1.0	0.1	3.0	
RENÈ N5	Bal.											
SC 16	Bal.	16.0	5.0	3.0		3.5		3.5	3.5			

The materials listed in the previous tables can be regrouped in six classes as it is shown in Table 6:
class A - solid solution stregthened alloys;
class B - Co base cast alloys;

class C - Ni base precipitation hardened wrought alloys;
class D - Ni base cast alloys;
class E - 1st generation single crystal alloys;
class F - 2nd generation single crystal alloys.

TABLE 6 - Different groups of superalloys

GROUP	PROCESS	ALLOYS
A	Ni & Co sheets	Hastelloy X, Haynes 186, Nimonic 86
B	Cast Co	X 40, X 45, FSX 414, ECY 768
C	Wrought Ni	Inconel X750, Nimonic 105, Udimet 500, Inconel 700, Nimonic 115, Udimet 520, Udimet 710, Udimet 720
D	Cast Ni	Udimet 500, Inconel 713 LC, Renè 77, IN 100, Renè 80, MAR-M-246, IN 738LC, MAR-M-247, MAR-M-002, Renè 125, GTD 11, IN 939, GTD 222
E	SC Ni	PWA 1480, Renè N4, SRR 99, RR 2000, AM1, CMSX-3, CMSX-6
F	SC Ni II	PWA 1484, CSM X4, Renè N5

The ability of the turbine components to resist the combined effects of pressure loads, centrifugal loads, vibration and temperature is predicted by design referring to three basic failure mechanisms: creep, low-cycle fatigue and high-cycle fatigue for which material behaviour data and models are necessary.

Materials creep strength can be expressed, at least in the preliminary design phase, by the Larson-Miller time (t) temperature (T) parameter, $P = (T/10^3)(20 + \log t)$ which yields a characteristic master curve for each material. Fig. 7 gives the bands which contain the typical master curves of materials belonging to the above defined six classes.

Taking into consideration blade materials (classes C, D, E, F), Fig. 7 gives a picture of the tremendous increase in creep resistance capability, mainly associated with process metallurgy developments.

Table 7 gives a summary of ranges for ultimate tensile strengths and elongations at 870°C. For material selection, as well as for the very preliminary design, fatigue behaviour can be described through the tensile characteristics according to the Manson model, by which the low cycle fatigue resistance in terms of strain range is related to the material tensile rupture elongation, whereas the high cycle fatigue resistance in terms of stress range is proportional to the ultimate tensile strength.

TABLE 7 - Tensile properties at 870°C

GROUP	UTS (MPa)	El. (%)
A	310 - 420	50 - 85
B	310 - 410	16 - 22
C	640 - 830	15 - 30
D	660 - 900	5 - 18
E	980 - 1020	10 - 20

The development from wrought to cast airfoil alloys has been detrimental as far as low cycle fatigue is concerned, nevertheless the subsequent evolution step from conventional casting to the directional solidification process has yielded an overall mechanical behaviour improvement. In Fig. 8, where the low cycle fatigue resistances of a conventionally cast and a single crystal material are compared, the superiority of the directionally solidified superalloys is evident.

Physical metallurgy

Chemical Composition Evolution

In 1930s Al and Ti additions to Ni-Cr solid solution produced γ' strengthening as it was later discovered by TEM. Larger amounts of γ' formers Al, Ti, with Nb and later Ta, and solid solution strengthening elements Co, W and Mo, completely substituting Fe, were introduced in high temperature *Ni-base superalloys* in 1950s when casting to shape was introduced for vanes and blades. Exaggerated substitution of strengthening elements for Cr led to hot corrosion sensitive compositions, hence Cr level was partially restored in 1960s. In smaller amount Zr, B and Hf were added while C was present from the beginning.

Materials process evolution often drastically influenced composition. Casting to shape allowed to increase the strengthening phase volume fraction over 50% in late 1950s and specially in 1960s. Later on, DS processing technique that evolved to single crystal technology, with first SC blades introduced in engine starting in 1980, deeply changed compositions of Ni-base superalloys by further increasing solid solution and γ' strengthening and wiping out grain boundary stabilizing elements, i.e. C, B and Zr. The latter novelty produced an increase of incipient melting temperature, allowing higher service temperature. Higher Al levels have been necessary not only for further increment of γ' volume fraction but also for better high temperature oxidation resistance.

Recently Re is beeing sistematically added in the last (second) generation of SC alloys due to its effectiveness as matrix strengthener and in order to slow down high temperature diffusivity, together with W, Ta and Mo, hence augmenting metallurgical stability in service.

Cobalt-base alloys underwent a less dramatic evolution starting from Cr and Ni in solid solution and containing high C levels to produce precipitation hardening, to stabilize grain structure and for better castability; more solid solution strengthening (W, Ta, Fe) and carbide strengthening has been added during the first decades of superalloy development.

Although of simpler composition and microstructure than Ni-base, Co-base alloys are widely used for vanes and other non rotating high temperature parts due to their good castability, thermal fatigue, hot corrosion and oxidation resistance and repairability.

Phases and Microstructures

It is important to remind that at extreme operating conditions, about 15% below the absolute temperature melting point, a superalloy is fastly changing its microstructure, is being degraded at surface interacting with gaseous environment or, sometimes, with coating phases; constitution phases and their composition can also change during service.

The face centered cubic γ matrix solid solution strengtheners of Ni- and Co-base superalloys are mainly Co (Ni), Cr, Mo, W and Ta, the latter three beeing also particularly effective high temperature stabilizers of microstructure.

Grain boundaries are pinned by carbides, borides and phases containing Zr.

The $L1_2$ coherent second phase γ' of Ni_3Al type is the main strengthening phase of Ni-base alloys, while it does not appear in Co-base alloys. The ordered structure is its main strengthening factor that is enhanced when temperature is increased. γ' forming elements are Al, Ti, Nb, Ta and W; they must be carefully balanced also to avoid large mismatches with matrix that would favour γ' microstructural instability.

Chromium and Al form a surface oxides film preventing further diffusion of oxidizing elements and action of corroding compounds. High Al concentration is essential for oxidation resistance over 1000°C, while Cr acts more effectively below that temperature.

Secondary carbides, i.e. $M_{23}C_6$ and M_6C, stabilize grain boundary structure of wrought and cast alloys, while MC primary carbides, copiously present in cast alloys, release C and γ' forming elements Ti, Nb and Ta during high temperature exposure.

Some non desired embrittling phases or phases that subtract useful elements to the alloy can also appear, as shown in Fig. 9, where a chronological evolution of superalloy microstructure is schematized: a continuous increase in γ' volume fraction and particle size is evident, with carbides more present in conventionally cast (CC) than in wrought alloys, less present in large oriented grain structures (DS) and finally disappearing in single crystals. Also γ-γ' eutectic nodules that abund in CC alloys disappear in advanced technology microstructures where large cuboidal γ' particles are present coalescing in lamellar raft-like structure during service. In Fig. 9 deleterious phases as η, Laves, σ, μ, sulfides and others also appear.

Since grain boundaries represent weak surfaces for creep and environment resistance, the evolution of processes brought about first larger grains with contorted boundaries (CC), than large oriented grains to minimize grain boundary surface transverse to principal stress and finally [001] oriented single crystals with high creep resistance and ductility, high thermal fatigue resistance due to low Young modulus and good response to oxidizing environment.

Cobalt superalloys creep strength depends mainly on solid solutes and carbides (MC, M_6C, $M_{23}C_6$,..) since γ' does not form. The microstructures and their evolution are less complex than in Ni-base alloys and the process technology has not been pushed to SC technology.

Present developments

Oxide dispersion strengthened (ODS) superalloys, intermetallic componds (IMC) and ceramics, both as monolithic materials or composites, including C/C composites shall be examined. Fig. 10 shows schematically how strengths of these material systems compare on the temperature scale with superalloys.

Eutectic alloys and Nb alloys are here shortly mentioned:
eutectic alloys have been shown to be technically viable but, due to the intrinsically slow production process their cost is too high against the modest advantages shown to compete with SC components; Nb alloys are the only refractory metal based alloys that, due to their not very high density and relative tolerance of oxidation environment, could be used for some special component application in non oxidizing environment when higher termperature capability is mandatory.

Oxide dispersion strengthened superalloys

In 1970s mechanical alloying (MA) followed by thermomechanical treatment and recrystallization heat treatment proved to be technically viable to produce sufficiently homogeneous ODS alloys having long recrystallized grains. Alloy MA 754, showing excellent thermal fatigue and high temperature mechanical resistance due to the high stability of its Y_2O_3 hardening dispersoid, has been adopted for vanes at least in one application. The γ' precipitation and Y_2O_3 strengthened alloy MA 6000 was designed to enhance intermediate temperature mechanical resistance. The Larson-Miller plot in Fig. 11 shows that the advantage of this alloy over SC blades is apparent at very high temperature, but unfortunately at too low stresses for blade applications. MA 6000 blades are often coated to reduce oxidation. Recently the more oxidation and sulfidation resistant alloy MA 760 containing higher Al and Cr levels has been proposed, although a european concerted action effort to use this ODS alloy for heavy duty gas turbine vanes has not been successful. Table 8 shows the composition of ODS alloys for structural high temperature applications.

TABLE 8 - Oxide dispersion strengthened alloys and their composition

ALLOY	Ni	Cr	Mo	W	Ta	Al	Ti	Fe	C	B	Zr	Y_2O_3
MA 754	Bal.	20.0				0.3	0.5	1.0	0.5			0.6
MA 6000	Bal.	15.0	2.0	4.0	2.0	4.5	2.5		0.05	0.01	0.15	1.1
MA 760	Bal.	20.0	2.0	3.5		6.0			0.06	0.012	0.15	1.0
PM 3030	Bal.	17.0	2.0	3.5	2.0	6.0			<0.05	0.01	0.15	1.1

Intermetallic compounds

After decades of study of high temperature long-range order IMC and an increasing effort in last 6 years in particular in USA, both monolithic IMC and IMC matrix composites are breaking on the scene and their use is predicted to increase significantly at the very beginning of next millenium. In general high specific stiffness, UTS and creep resistance are the advantages, while brittleness and poor toughness at room temperature and difficult fabricability are the negative aspects of these compounds.

Titanium and Ni aluminides are the most promising IMC for gas turbine applications and, in particular the γ TiAl-base system is the most attractive in end stages applications, as illustrated in Fig. 12.

TiAl-base compositions of materials in development range between Ti/46-52 at % Al/1-10 at % X, where X is one or more of the following elements: V, Cr, Mn, W, Mo, Nb or Ta. A two phase γ alloy or a three phase $(2\gamma+\alpha_2)$ alloy are preferable for high temperature strength. The limited oxidation resistance, rather than the mechanical strength, restricts the high temperature use for TiAl-base materials requiring coating for fuller exploitation.

The low ductility from which aluminides suffer can be improved by refining the grain size. Fine grain size can be obtained with hot isothermal forging, extrusion or powder processing of TiAl, while NiAl requires powder processing followed by extrusion. NiAl-base IMC have excellent oxidation resistance and low density compared to superalloys. The development of a multiphase NiAl-base material, specially if used as fibre-composite matrix, with acceptable room temperature toughness and high temperature metallurgical stability at reasonable cost is an important challange.

The National AeroSpace Plane (NASP) is the prime motivator for R & D work on IMC in USA. Japan launched in 1989 an 8 years national effort mainly on Ti and Nb aluminides. Efforts for european concerted actions (Brite/Euram and COST) are being spent only recently, while Germany already launched a national programme.

Ceramics

Monolithic ceramics have been extensively studied in the 1970s and early 1980s in research programmes aimed to develop a ceramic gas turbine for automotive propulsion. The interest has been focused on silicon carbide, silicon nitride and zirconia. Several processing routes have been established and patented. A major problem with ceramics is the inability to manufacture them with reproducible characteristics, a problem mainly related to the volume effect on the allowable tensile stress level and to the fabrication internal defect level. Another handicap is the poor fracture toughness since flaws caused by damage during service can lead to reduction in strength and catastrophic failure.

Ceramic materials are the thermal barrier coatings currently sprayed onto the surface of combustor liners and nozzle guide vanes platforms. The preferred ceramic is zirconia because of its very low thermal conductivity and its relatively high coefficient of thermal expansion (among ceramics the closest to typical Ni-superalloy coefficient values).

The application of thermal barriers to airfoils is in an advanced stage of development. Thermal barriers typically allow a metal temperature decrease of about 60 K or a cooling flow reduction of about 25%.

It is unlikely that components entirely in ceramic will be introduced unless continuously reinforced. A ceramic composite developed in Europe is silicon carbide fiber reinforced silicon carbide (SiC-SiC). The main problem related to SiC/SiC is that the commercially available fibers do not maintain their strength above 1100°C. Components have been tested in engine environments with particular reference to reheat applications.

Structural carbon materials in the form of carbon-fiber-reinforced carbon matrix composites (C/C) have an excellent mechanical performance. These composites are presently used for space propulsion. Poor oxidation resistance above 500°C is the main barrier to moving C/C to aeropropulsion applications. As a result, the development of reliable oxidation protection is crucial. The current state of the art in protection of high performance carbon-carbon composites is based on the use of silicon carbide or silicon nitride as primary oxygen barriers, coupled with internal inhibitors and sealants. Since carbon materials have a substantially lower thermal expansion coefficient than the coating material, and the coating is deposited at elevated temperature, cracks form and open during cooling, whereas they close mechanically at high temperature. The sealants design has a key role to make effective coatings by sealing thermal stress cracks.

Both the two high temperature composites, C/C and SiC/SiC, are candidates for turbine applications. The efforts put on their development depend on the preference in considering the possible enhancements, in coating effectiveness for C/C and in fiber stability for SiC/SiC. Fig. 13 shows typical specific stength data of both the composites as well as the data of a monolithic silicon nitride ceramic, compared to a single crystal superalloy.

Final considerations

At present exclusively nickel and cobalt superalloys are used in gas turbine hot parts. The development leading edge is constituted by advanced materials for aeroengine applications, where the superalloys potentiality has been almost entirely exploited.

ODS superalloys seem the ultimate possible development but their application is not considered effective at present. Despite the fact that ODS materials are candidate for combustion chamber parts as well as for airfoils, their lack of adequate forming and welding capability will require new fabrication techniques. The same techniques could be applied to enlarge the use of SC alloys to components with threedimensional geometry like nozzle guide vanes. In other words the chance of a possible further development of superalloys depends on fabrication techniques development more than on new materials properties.

Weight of components is of prime importance in aeroengine applications, hence big efforts are devoted to the development of light materials, like titanium aluminides and ceramics. In this case an obstacle to a short term application is also the lack of adequate stress and lifing criteria, since the nature of these new materials does not allow to extrapolate design criteria from the experience on conventional materials.

Time to application could be shortened by innovative concepts in terms of component design, for example design approaches able to realize a compression stress state in ceramic airfoils.

Although the aeroengine materials research is the leading edge, the challenging requirements put by the industrial gas turbine development needs must not be underestimated. In the latter case: i) the corrosion resistance in low grade fuel combustion environment, ii) the design requirement of very long service lives, and iii) the ability to cast large airfoils with DS microstructure are peculiar features which make superalloys further developments still possible. Besides, industrial gas turbine application is driving the protective coatings development aimed to improve corrosion resistance and component temperature capability.

In conclusion a possible scenario of future developments progression is the following:

— extension of DS casting technology to large components like industrial gas turbine blades;

— introduction of new fabrication techniques able to allow the use of SC alloys for components with complex geometry;

— application of thermal barrier coatings to aeroengine and industrial gas turbine airfoils;

— introduction of titanium aluminides in turbine end stages;

— introduction of monolithic and composite ceramics for static hot parts and subsequently for rotating hot parts.

Last but not least, considering the spectrum of new materials under development, the management role is critically important in selecting among: i) competing technologies, ii) potential options of cooperative programs, and iii) research paths able to shorten time to application.

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[11] *Materiali per parti calde di turbine a gas in applicazioni terrestri* (Atti Giornata di Studio, AIM, Milano, 1991).

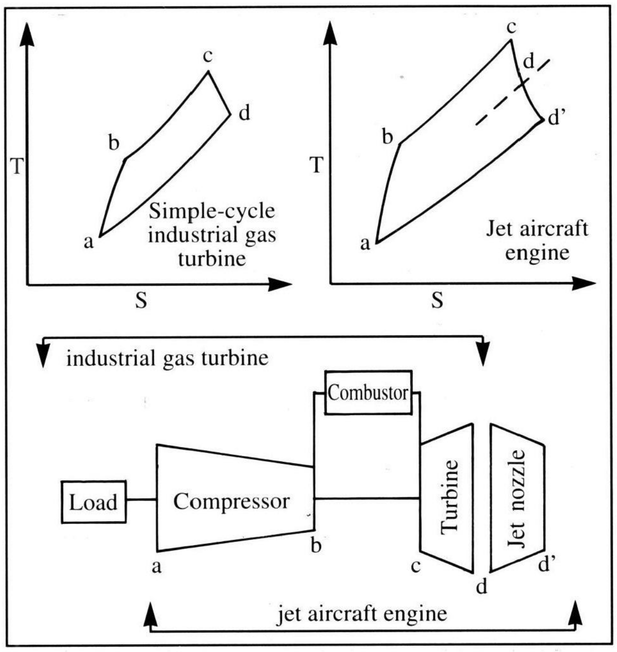

Fig. 1:
Turbine engine diagram.

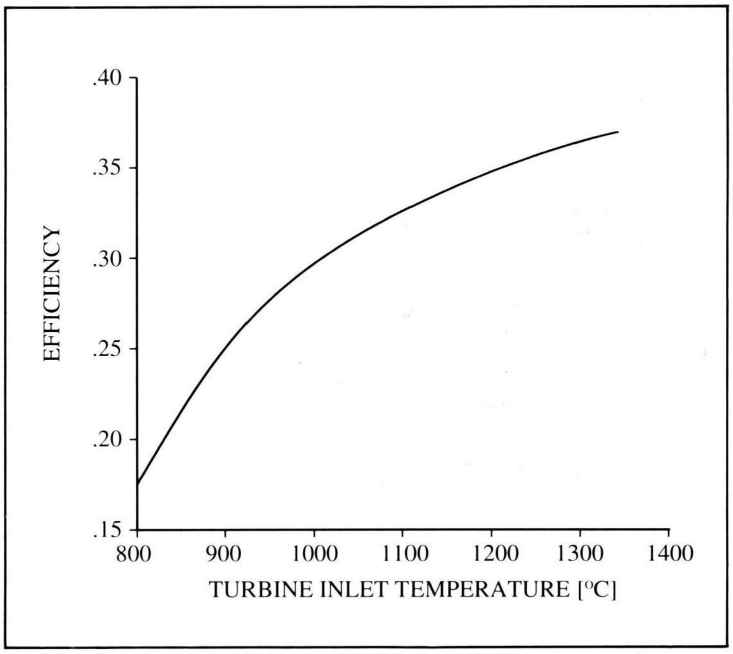

Fig. 2:
Example of temperature influence on efficiency of an industrial gas turbine.

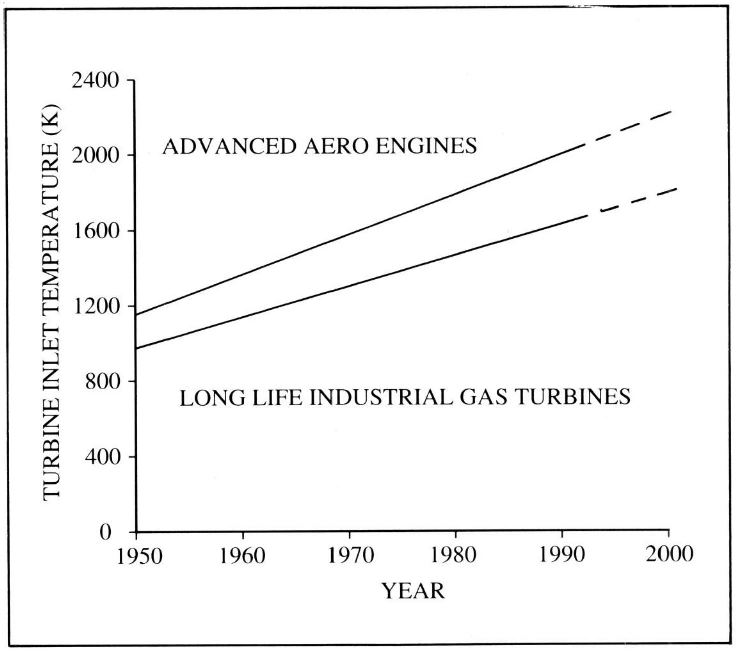

Fig. 3:
Turbine inlet temperature trend in aeroengines and industrial gas turbines.

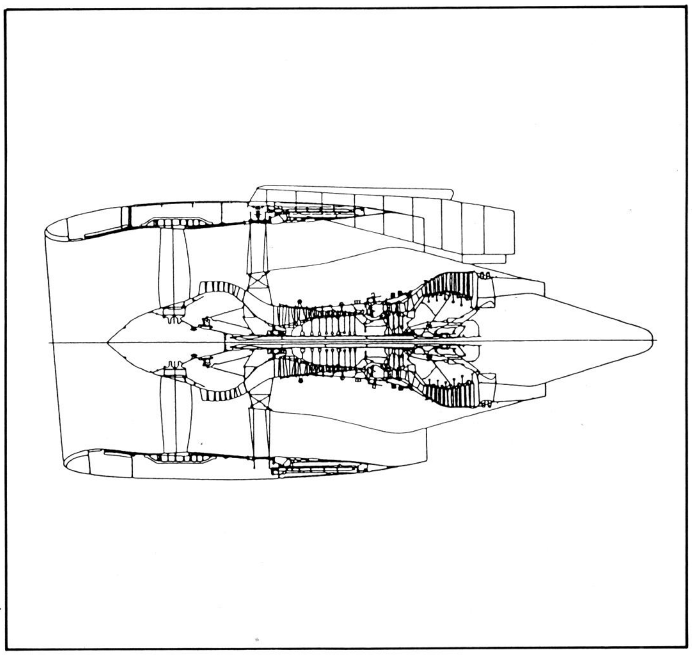

Fig. 4:
GE 90 aeroengine.

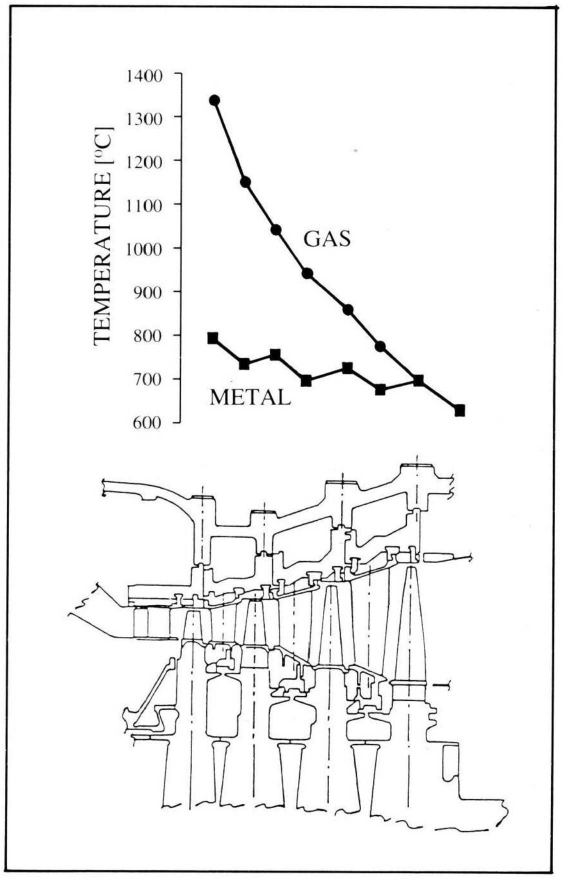

Fig. 5:
Turbine section of the industrial gas turbine FMW 701F.

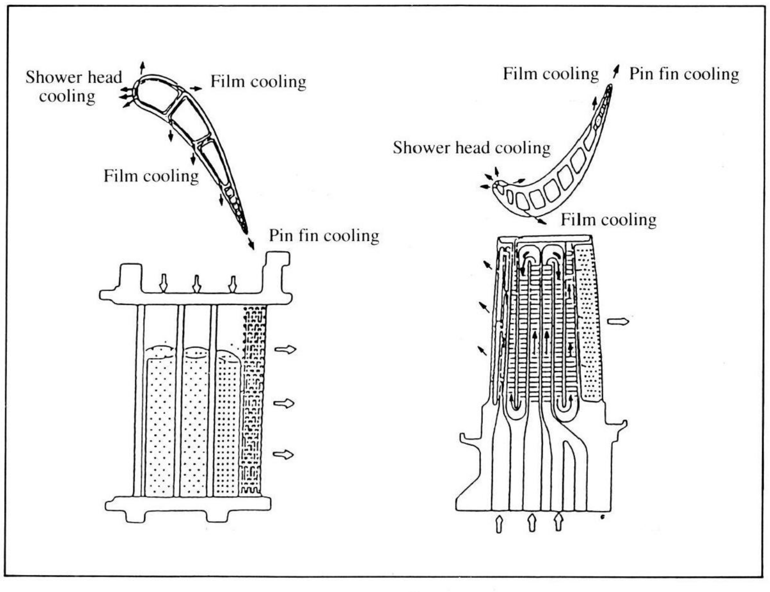

Fig. 6:
FMW 701F first stage airfoils geometry: nozzle guide vane and blade.

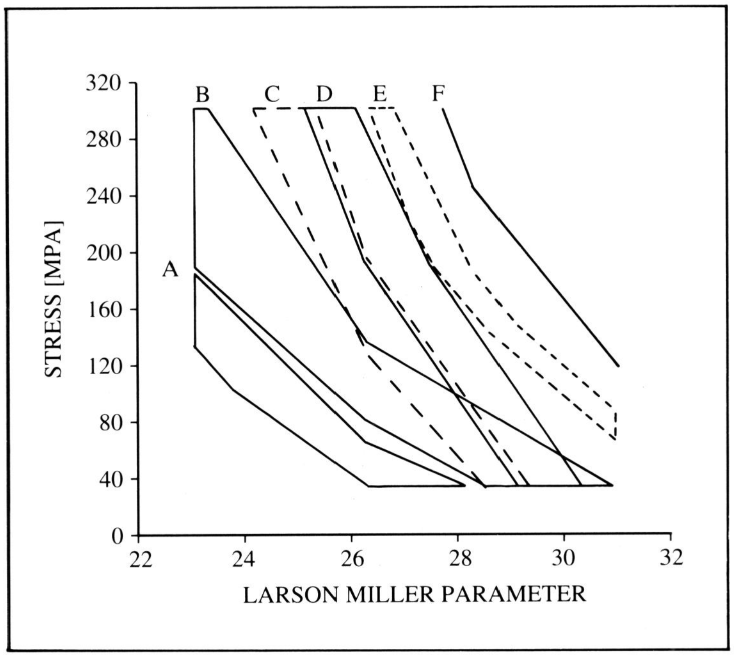

Fig. 7:
Creep strength in terms of the Larson-Miller parameter (temperature in K and time in h).

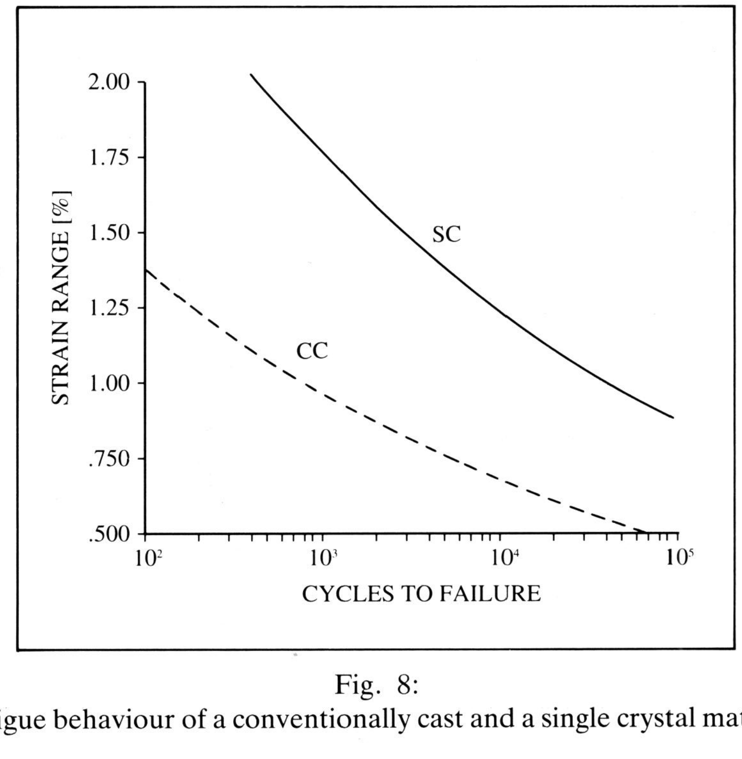

Fig. 8:
Low cycle fatigue behaviour of a conventionally cast and a single crystal material at 750°C.

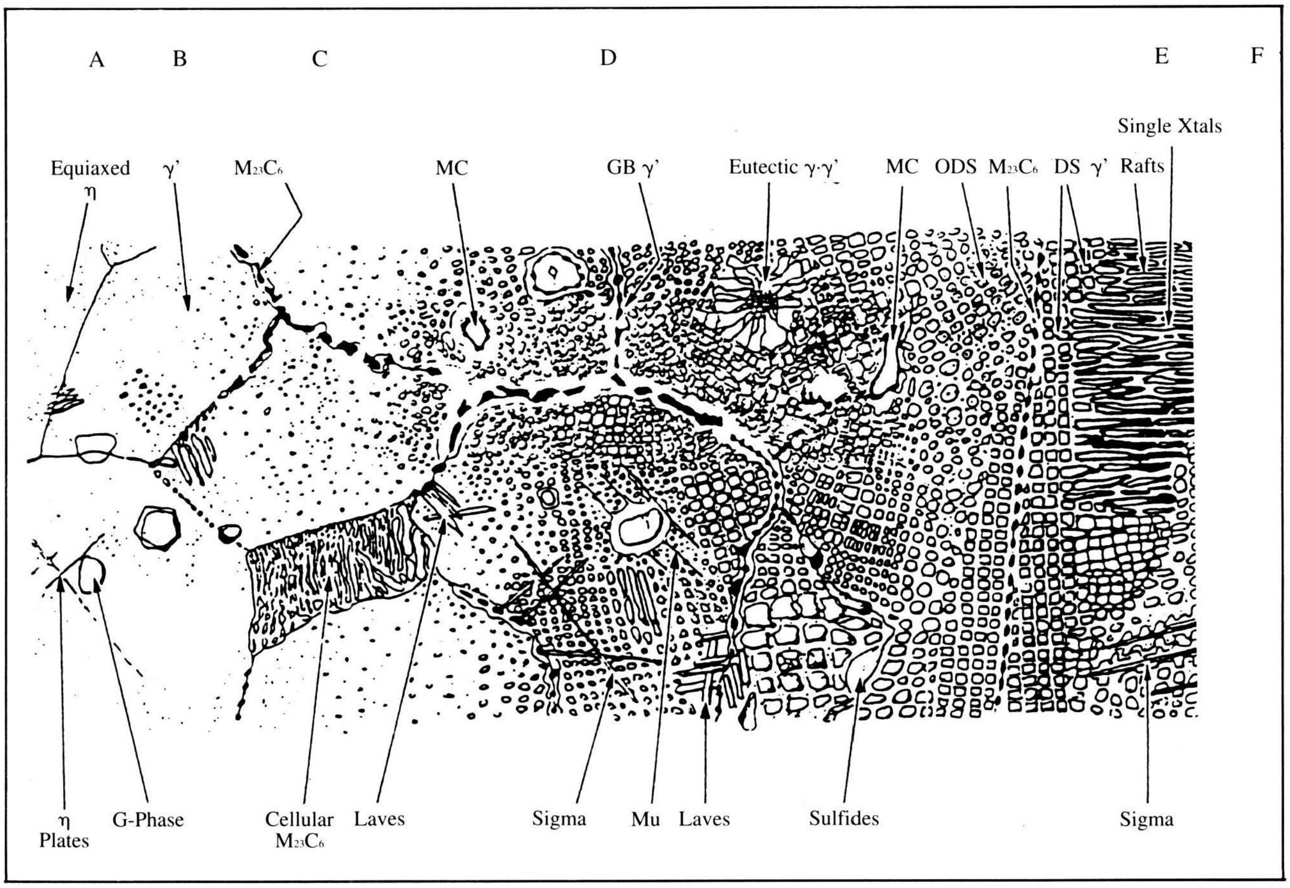

Fig. 9:
Representation of the chronological evolution of superalloy microstructure[5]. A-F represent the alloys groups reported in Table 6.

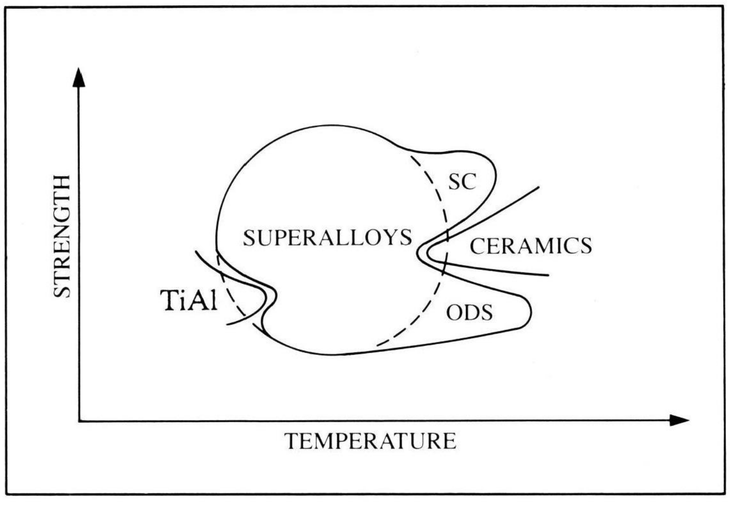

Fig. 10:
Qualitative comparison of the strength of various materials systems on the temperature scale.

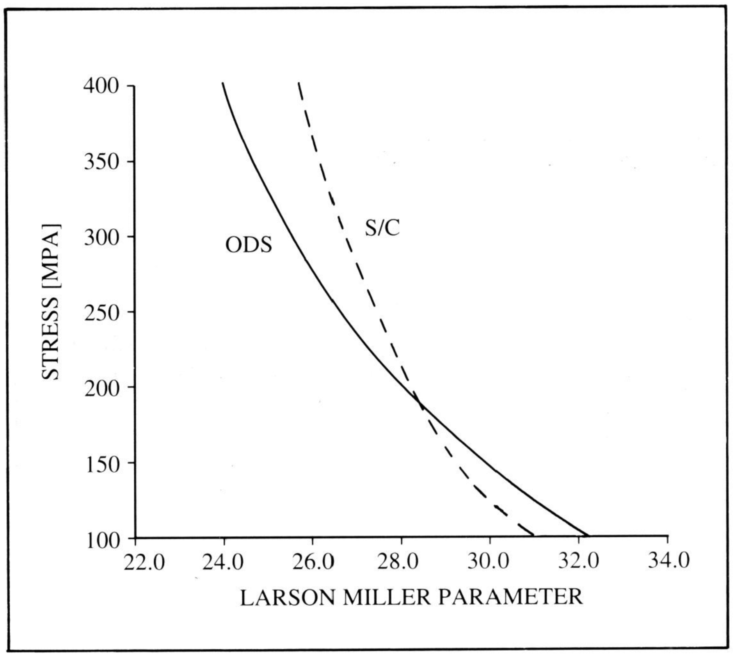

Fig. 11:
Larson-Miller graph showing better creep resistance of MA 6000 (ODS) compared to a first generation SC material at very high temperature.

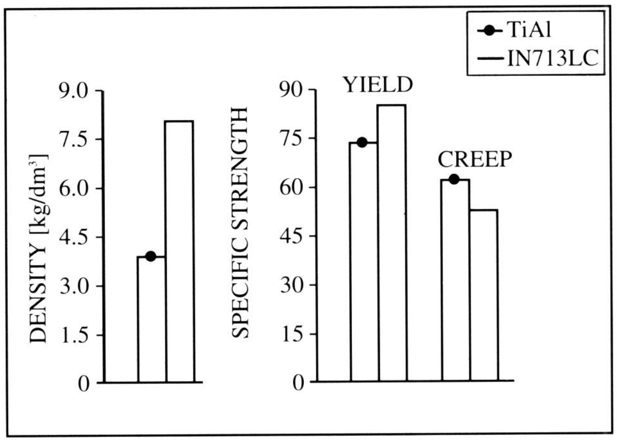

Fig. 12:
Density (kg/dm^3) and specific strength (MPa dm^3/kg) comparison of a CC material with an IMC material (crept at 800°C/100h).

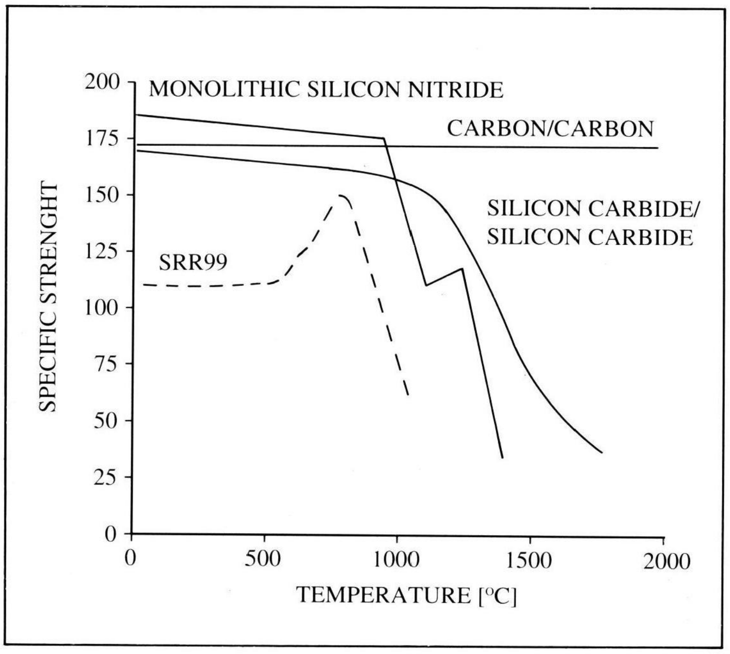

Fig. 13:
Comparison of present blade material specific strength with non-metallic materials under development.