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Materials Review and database

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WP2.1 High temperature materials for internals

The aim of this task is to establish a material database on possible materials for HTR control rods. The objective is to select a material with superior high temperature properties, compared to metallic alloys.

In a first attempt to clarify the needs for such materials, we give here a short overview of R&D work performed at JAERI on this subject. In the last five years, JAERI has performed several tests on Carbon/Carbon (C/C) composites for the control rods of the HTTR test reactor [JAERI 96, JAERI 98].

The most recent article found on this subject is given as a separate ".pdf" file (JAERI-Research 98-003). It contains data on the two grades of C/C composites evaluated at JAERI.

The materials for the control rods of the HTTR are required to be stable at very high temperatures. Heat resisting alloy, Alloy 800H, was selected for this application. Due to the temperature limitation of this alloy, the control rods are inserted in 2 steps to avoid damage by high temperatures (>800°C) and heavy irradiation in the core of HTTR. Nine out of the 16 pairs of control rods in the reflector region are inserted immediately at the time of scram, while the other 7 pairs in the core region are to be inserted 40 minutes later when the temperature of coolant at the outlet becomes lower than 750°C. Table 1 summarises the conditions in which the control rods are operated in the HTTR reactor.

	Maximum temperature (℃) (Bottom of the lowest sleeve)		Fluence (660 days***) (Bottom of the lowest sleeve)
	Normal operation (950℃) (Location : B+0.75 level)	Scram (Loss of commecial electricity)	
Central control rod	~610*	~1050** (One-step insertion) ~800* (Two-step insertion)	~6×10^{24} n/m^2
R2 control rod	~560*	~900*	~6×10^{24} n/m^2

* : Systematic and random factors conidered.
** : Temperature of control rod guide block.
*** : Lifetime of the control rod is about 5 years (depends of creep property).
B : Top of the core.

Table 1 : Service conditions for the HTTR control rods [Jaeri 98].

[2] Part of this work on turbine materials results from a paper written with C. Escaravage (Framatome). This paper can be found on http://www.iaea.org/inis/aws/htgr/index.html.

Carbon/Carbon composite control rods are therefore very attractive materials for high temperature use (1000°C) as the reactor core would operate with an increased margin of safety.

After first R&D work carried out on the base materials ([JAERI 96]), JAERI succeeded in making C/C composite control rods with the concept shown in figure 1. This control rod consists of an outer and inner sheath tube, bolts and joints, which are all made of three dimensional C/C composite material. New three dimensional carbon fibre weaving techniques were applied in the manufacture to enforce the structure. In particular, the rod is strengthened in circumferential, radial and axial directions with a high content of carbon fibres, and matrix material is filled with a specific process to achieve a high density.

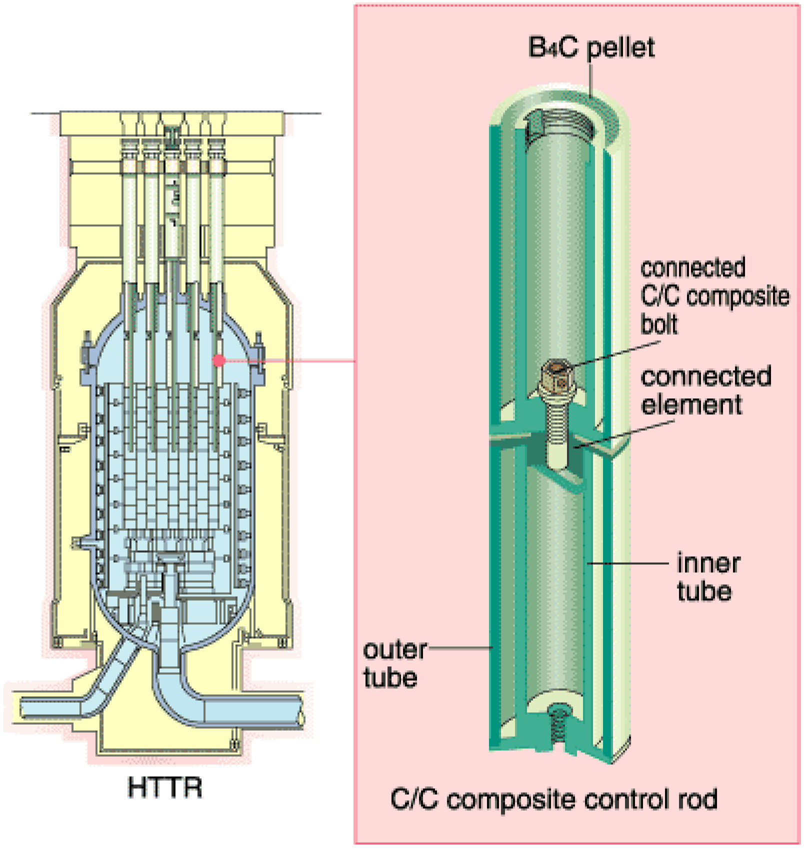

Figure 1 : Conceptual structure of the HTTR and the C/C composite control rod.

Mechanical tests on C/C tubes showed a stable fracture behaviour and a superior strength, when compared to metallic alloys. In a recent paper, JAERI has summarized the data requirements for the design of the control rods (table 2).

Test item	Properties	Database
Strength test Fatigue test	Static strength, Fracture strain, Strength distribution, Fatigue strength, Effect of environment and irradiation	Design stress strength, Design yield stress, Design tensile strength, Poisson's ratio, Longitudinal elastic modulus, Allowable fatigue life
Creep test	Creep strength, Creep fatigue	Design creep fracture strength
Thermal expansion		Thermal expansion coefficient
Thermal conductivity		Thermal conductivity

Table 2 : Necessary material testing and database for the design of the control rods [JAERI 98]

WP2.2 High temperature materials for the gas turbine

The aim of this task is the selection of metallic materials for the turbine disks and blades of a direct-cycle high temperature gas reactor.

Recent advances in the design of turbomachinery, recuperators and magnetic bearings provide the potential for the use of the High Temperature Gas Reactor (HTGR) with a closed cycle gas turbine. The reactor size has been reduced in developing the passively safe module design and the size of industrial gas turbine has increased to accommodate the energy released from a HTGR module. Highly effective compact recuperators have been developed, they are a key requirement for achieving high efficiency. The availability of large magnetic bearings has also eliminated the potential problem of coolant contamination by the oil of lubricated bearings.

National and international R&D programs are under way to explore areas where technical development is needed [IAEA 96]. Among these programs, the closed-cycle gas turbine concept is investigated in several industrial projects [IAEA 00].

- The Pebble Bed Modular reactor (PBMR) in South Africa, with a gross electrical generation of 117 MWe. The project schedule has been established, with deployment of a first unit in 2006.

- The Gas Turbine Modular Helium Reactor (GT-MHR) developed by an international consortium, with a targeted 286 MWe generation per module, a prototype single unit is scheduled in Russia in 2009.

- Test Reactors are being commissioned in Japan (High Temperature Test Reactor HTTR) and China (HTR10) to evaluate the safety and performance of HTGR. These reactors are designed with an indirect conversion cycle, but they will support R&D activities like electricity generation via the gas turbine and high temperature process heat applications.

The development of advanced HTGR concepts requires materials data and understanding of materials behaviour under reactor operating conditions and environment. The components for the primary circuit operate at temperatures above 600°C and up to 850°C in order to reach high efficiencies. For these materials and components, considerable data and know-how exists, but industrial and technical feasibility are not well established for the most recent designs. For example, a significant effort is needed for the turbine components, for which the expected working conditions are beyond todays industrial capabilities. The fundamental technologies required for the design of the turbomachine have been proven for aerospace and industrial gas turbines. However, the HTGR gas turbine requires a design that includes high temperature capability and long term endurance. The required design properties, associated with the large sizes of components, can be seen as a major issue.

In this work, a review with a selection of most relevant materials and processes is presented, for both turbine blades and disks. The focus of this work is to discuss the choice of high temperature materials in accordance with the recent HTGR design specifications. The GT-MHR turbine design is used as a reference for high temperature materials selection. Also an evaluation of recent advances in materials for industrial gas turbine is presented, with a brief overview of associated fabrication processes.

1. HTGR TURBINE DESIGN SPECIFICATIONS.

The GT-MHR module consists of a nuclear source of heat (reactor system) and the Power Conversion System (figure 2).

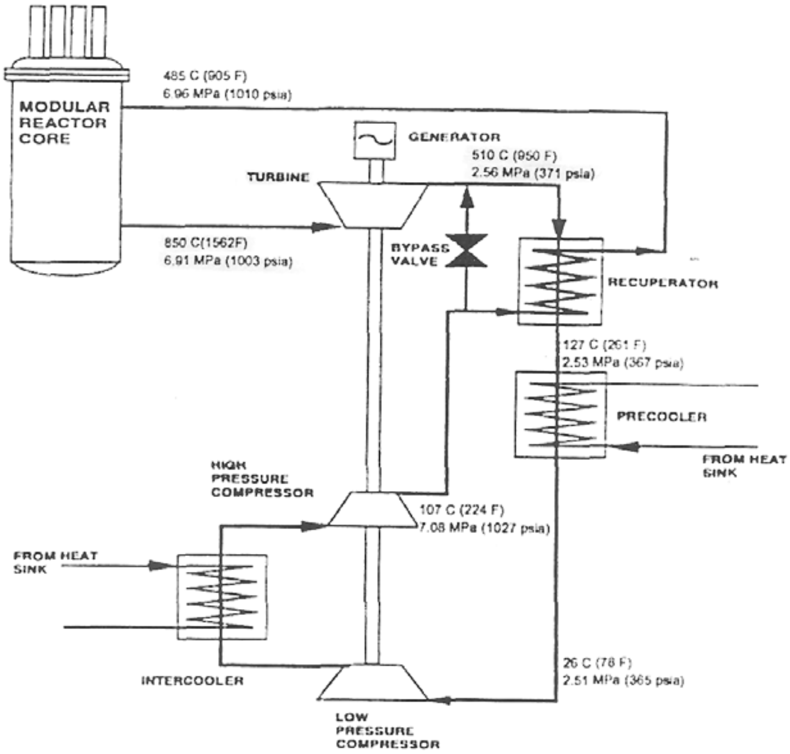

Figure 2 : GT-MHR flow diagram [IAEA 2000].

The Power Conversion System (PCS) receives approximately 600 MW of thermal energy from the reactor system and converts 286 MW of net usable electrical energy, with an overall efficiency of approximately 47%. The PCS includes the turbomachine, the precooler and intercooler, and connecting pipelines. The turbomachine consists of the turbocompressor, the electrical generator, bearings and seals. The turbocompressor includes the turbine and two compressor sections (low pressure and high pressure compressor).

The helium is received by the PCS from the hot gas duct. The helium then expands through the gas turbine which is coupled to the electrical generator. From the turbine exhaust, the helium flows through the hot side of the recuperator, transferring heat to helium returning to the reactor. The helium leaving the hot side of the recuperator is cooled (precooler) before passing through the low-pressure compressor, intercooler and high-pressure compressor. The helium then passes in the cold side of the recuperator where it is heated for return to the reactor system.

The entire assembly is installed in a vertical orientation and is rotating at 3000 rpm (50Hz). The turbine design is based on the technology available for large industrial engines. Substitution of helium for air in this nuclear gas turbine modifies aerodynamic requirements by removing Mach number limitations. As the rotational speed is fixed (synchronous generator), the size of the turbine is dictated by the choice of blade speed. The highest blade speed is desirable to limit the number of stages, but it is limited by the high temperature stress limits for the blades. In the GT-MHR concept, a 12 stage turbine with very high efficiency was established. With a turbine inlet temperature of 850°C, blade and disks cooling are not considered as necessary in the design. Salient features of the GT-MHR turbine are given in table 3.

Power, MW	560
Rotational speed, rpm	3000
number of stages	12
Expansion coefficient	2.67
Adiabatic efficiency, %	93
hub diameter, mm	1400
Inlet parameters	
temperature, °C	848
pressure, MPa	7.0
He mass flow, kg/s	320
pressure losses, MPa	0.007
blade height, mm	170
Outlet parameters	
temperature, °C	510
pressure, MPa	2.62
pressure losses, MPa	0.02
blade height, mm	225

Table 3 : GT-MHR turbine characteristics at full power [Couturier 00].

2. CRITERIA FOR MATERIAL SELECTION.

For the material selection, the key points are the first stage turbine disks and blades where the stresses and temperatures are the highest. The materials for these components should ensure a safe operation for at least 60000 hours. Therefore, a ground rule for the material selection of the most critical parts was proposed as following [Couturier 00].

- for the blade alloy $\sigma_{60000h}^{850°C} \geq 225MPa$
- for the disk alloy $\sigma_{60000h}^{850°C} \geq 180MPa$

The use of high temperature materials for nuclear systems will require an extension of material design rules and codes currently used. From the designer side, increasing the temperatures means a change from time-independent to time-dependent characteristics (creep, creep-fatigue).

Figure 3 illustrates the maximum temperature of metallic components in current and future reactor systems. In addition, figure 3 indicates the maximum temperature considered in the design codes. Although some existing codes use time-dependent properties in design (ASME Code Case N47 now incorporated as subsection NH of section III and RCC-MR), the temperature limits of materials included in the design codes available today are significantly lower compared to HTGR needs and therefore a significant step forward is required. The first criteria being high temperature tensile strength and stress to rupture, other properties, such as resistance to thermal fatigue must be considered in design. Notch sensitivity is also probably detrimental to the high cycle fatigue of rotating parts of complicated shapes.

Possible corrosion effects of environment should also be considered for the material choice as helium impurities can influence the metallic material properties in service via carburisation and oxidation mechanisms. However, for the complete assessment of structural integrity, the combined effect of different failure modes must be considered and particularly the following.

- Creep-fatigue :- it is generally considered that the amount of fatigue endurance reduction by creep or relaxation is increased when the material has low creep strength. Creep ductility can also be considered as a limiting parameter in creep-fatigue.

- Fatigue and environment effect :- initiation of cracks particularly in case of thermal fatigue takes place at the surface of the material and can be influenced by the effects of the environment on the material surface.

- Creep-fatigue and environment effect :- the effect of environment can be important in creep-fatigue and in thermal fatigue, and experiments have shown that the reduction of fatigue life by hold times in relaxation can be quite different in vacuum from what is observed in air. The corresponding mechanisms are difficult to clarify for a quantitative prediction, the helium effects being suspected to be intermediate between vacuum and air effects.

As long term service is needed, long term stability of material properties must be assessed. When the optimum strength is obtained by a heat treatment at temperature lower or nearly equal to the service temperatures, there are some doubts about the stability of the corresponding microstructure.

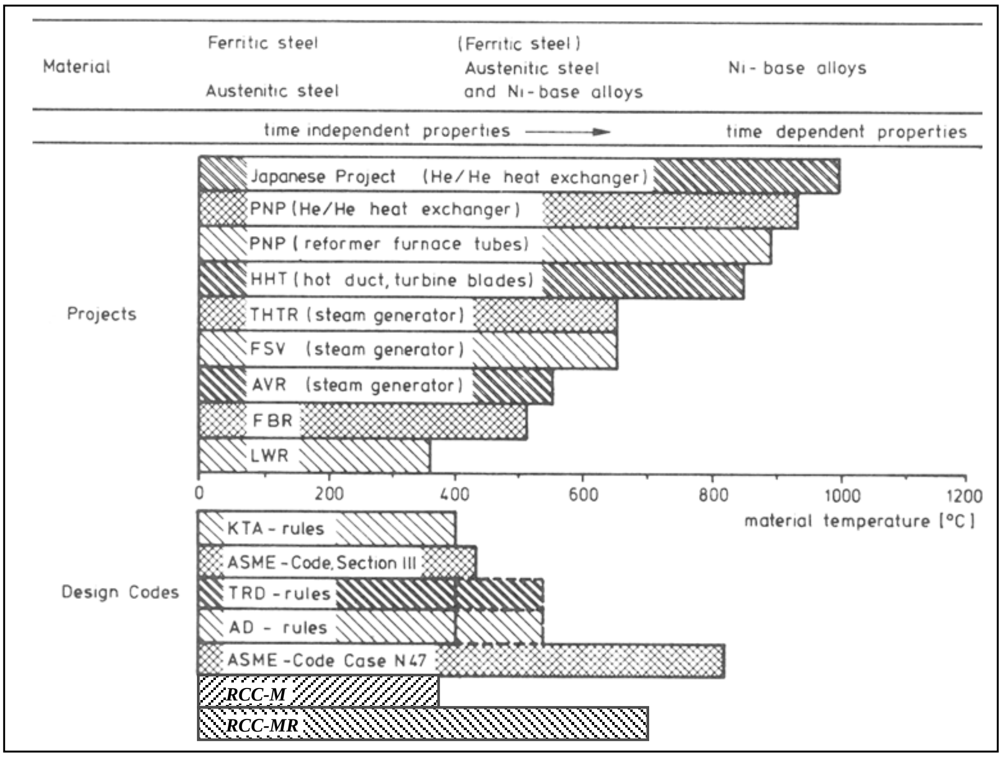

Figure 3: Material temperature and design codes for metallic components in nuclear applications (partly from [Nickel 84]).

Based on these requirements, the selection of metallic materials for blades and disks is discussed in the next sections. The chemical compositions of alloys that are evaluated is given in appendix 1.

3. MATERIALS FOR THE TURBINE DISKS.

At this stage, it remains unclear if alloys containing substantial additions of cobalt or tantalum would be selected for use in the turbine of a HTGR. In some past studies, alloys containing more than residual levels of these elements have often been excluded from the list of reference alloys for the primary circuit, because of potential radioactive contamination. However, many of the high temperature alloys with Cobalt have the highest mechanical properties and they are examined in the case of further acceptability [Graham 76, Couturier 00].

Inconel alloy 718 was selected for the HTGR turbine disks in recent studies. Alloy IN718 is a nickel-based precipitation hardened material. It has the necessary strength, short term creep and corrosion resistance only in the case of active cooling of the disk. With a turbine inlet temperature of 850°C, IN718 will require cooling to lower the temperature of the disks to around 650°C. Alternatively, alloys containing cobalt may be better in terms of high temperature strength (like Udimet 720 with $\sigma_{1000h}^{850°C} = 245MPa$).

For the GTMHR design stresses, classical nickel-base superalloys used for industrial gas turbines (IN718, Waspaloy) do not achieve the long term stress level requirements. The only existing alloy that could be envisaged for a non-cooled first stage disk is the alloy MA

6000. MA 6000 is an Oxide Dispersion Strengthened (ODS) superalloy that offers very high strength and microstructural stability. The creep strength of this grade could reach values higher than 185 MPa for 60000 hours at 850°C. This grade is produced by mechanical alloying of powders and subsequent Hot Isostatic Pressing (HIP) or hot extrusion [Benn 88]. The main difficulties with using ODS grades are that the process route is not mature and that much work is needed to produce large ingots with homogeneous mechanical properties.

In the frame of the Dragon project, Graham has proposed A286 and IN706 as reference grades for the disk production. These are Fe-Ni-Cr precipitation hardened grades with sufficient forgeability for large disk production but medium high temperature properties. Again, these grades can only be envisaged in the case of cooled turbine disks [Graham 76].

In the case of large components such as the GT-MHR turbine disks, the manufacturing capability is closely related to the strength of the alloy. In the case of nickel-base superalloys, the two major issues are the following.

- To obtain large ingots (~5-10 tons) without solidification porosities and macro-segregations. For recent superalloys, the manufacturing route includes a vacuum induction melting, a vacuum arc remelting and/or electro-slag remelting. The powder metallurgy would allow the production of high quality ingots with highly alloyed grades. In this case, microstructural inhomogeneities would be limited to the size of the powder particle.

- To forge these ingots which usually offer a very low hot-workability. Isothermal forging is used, but with the most recent superalloys (like Udimet 720), the maximum forgeable size is much lower than the GT-MHR disk diameters. Again, elaboration of near-net shape components by Hot Isostatic Pressing of powders appears promising for these large disks, as HIPping furnaces with large diameters are available (~1.4 m).

4. MATERIALS FOR TURBINE BLADES.

Materials selection and testing for HTGR turbine blades has been extensively studied. Essentially 2 types of metallic materials were investigated.

- Nickel-base cast superalloys.
- Molybdenum-base grades.

4.1. Ni-base cast superalloys.

In several past R&D programs, selected alloys were ranked by their high temperature strength, castability and cobalt content. The most promising alloys for the blades are therefore:

- 713LC [Nickel 84, Graham 76, Johnson 84, Weisbrodt 96, Schuster 76, Jakobeit 84];
- M21 [Graham 76, Johnson 84, Schuster 76];
- MAR-M 004 [Graham 76, Schuster 76, Jakobeit 84].

Alloy 713LC is a cast nickel base precipitation hardened alloy that combines superior castability and creep resistance. Alloy 713LC has the advantage of wide industrial experience in conventional gas turbine (turbine housings, case, stator). Alloy 713LC neither contains Co nor Ta and should therefore not present any contamination problems. In HTGR environment, alloy 713LC can be susceptible to carburisation and sulfidation problems, and coatings have to be envisaged. Alloy 713LC is well suited for the turbine blade required specifications, except for the first row of blades where cooling would be necessary to achieve the required lifetime. Directionally Solidified (DS) or Single Crystal (SC) blades would solve this problem, but alloys commonly used for DS or SC blades contain approximately 10% cobalt (DS Mar M 247, SC René N4).

Alloy 713LC was used for the fabrication of the turbine blades in the HHV project [Weisbrodt 96]. During this project held in Germany, a full-scale test turbine driven by HTGR-type helium was built. During the short high temperature lifetime experienced (~325 hours at 850°C), no material problems appeared with the working blades.

Alloy M21 is a low chromium nickel base alloy that combines precipitation hardening and solid solution hardening (~10% tungsten addition). It was selected for its superior corrosion resistance in impure helium.

Alloy MM004 has been developed from alloy 713LC and is claimed to have a better toughness because of a Hafnium addition [Jakobeit 84].

Among alloys that were also studied for a potential use as HTRG turbine blades, there are:

- IN 100 grade with good corrosion resistance [Johnson 84]. IN 100 is the reference alloy for PBMR power-turbine cooled blades [IAEA 96];
- Nimonic 80A with medium creep rupture strength which is envisaged for the last row of blades (lower temperature) [Nickel 84, Jakobeit 84];
- alloys Nimonic 90, 105 and 115, Udimet 520 and 700, IN591, IN738, IN792, René 80, M22, MAR-M 247, Nx 188 were also proposed in various studies [Graham 76, Jakobeit 84, Shainian 86]. FIS 145 alloy was developed in Germany for HTGR applications. It combines precipitation and solid solution hardening (~13.5% tungsten) with a high castability [Jakobeit 84].

In figure 4, a comparison of several blade alloys properties is shown, for creep rupture performed in both air and helium. Jakobeit pointed out the sharp decrease in creep strength for long times at 850°C, except for the Mo-base alloy. This decrease is less pronounced at lower temperatures, supporting the idea of a design with cooled blades to increase the blade lifetime [Jakobeit 84]. L-TZM is the strongest alloy in figure 4, this Mo-base grade is presented in the next section.

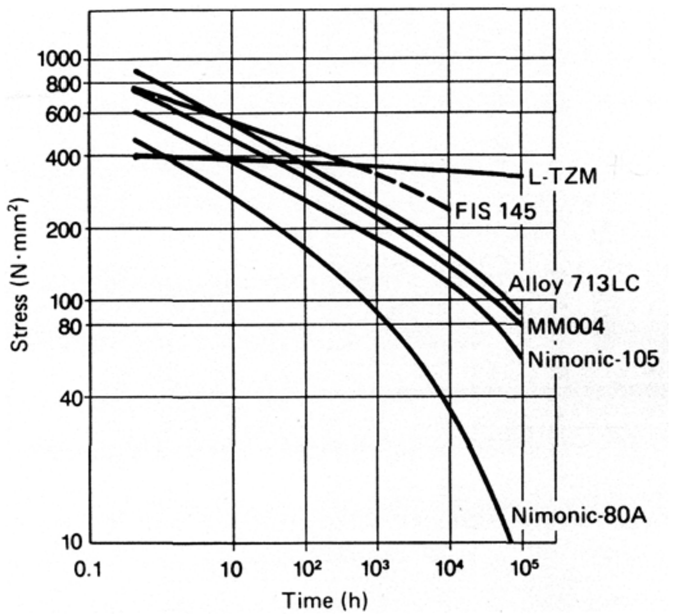

Figure 4: Creep rupture strength of different turbine blade materials at 850°C [Jakobeit 84].

4.2. Molybdenum-base alloys.

The molybdenum-base alloy TZM (Mo-0.5Ti-0.08Zr) has not been used in industrial gas turbines because of its poor oxidation resistance in air. However, R&D efforts performed in German HTGR programs have demonstrated a promising application for this alloy for helium turbine blades. As shown in figure 4, Mo-TZM exhibits a completely different creep-resistance behaviour as compared to the nickel-base alloys, mainly because of its high melting point (2607°C) [Jakobeit 83]. Almost flat creep curves make a 100 000h lifetime appear possible with an uncooled blade, as shown in figure 5 [Schuster 76].

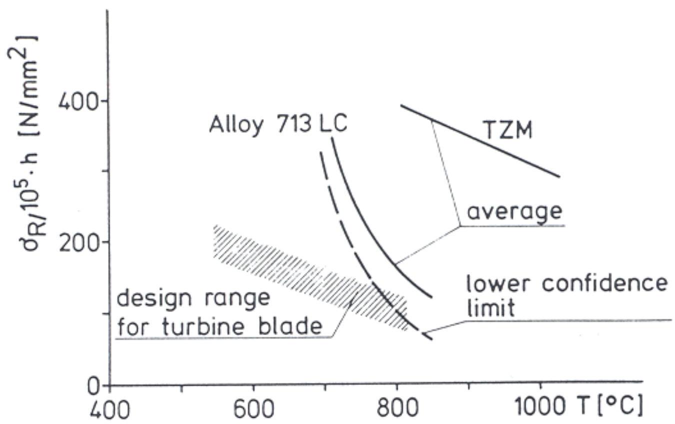

Figure 5 : Stress for rupture in 10^5 h versus design temperature for HTGR turbine blade [Schuster 76].

Precision forged blades were fabricated, starting from vacuum-arc-melted or powder metallurgy ingots. These blades showed a very high corrosion resistance in impure helium, but several fabrication difficulties and inherent drawbacks were encountered, and efforts for further development are almost abandoned today.

- Mo-TZM is brittle at temperatures up to 360-440°C and the notch sensitivity remains high at HTGR service temperature (figure 6).
- In case of air ingress in the circuit , this alloy will suffer a massive oxidation.
- Microstructure is heterogeneous in the blade, as a result of locally non uniform deformation state during hot forging. Low strength areas were found in fatigue and creep, especially in the airfoil/root transition zone [Jakobeit 83].

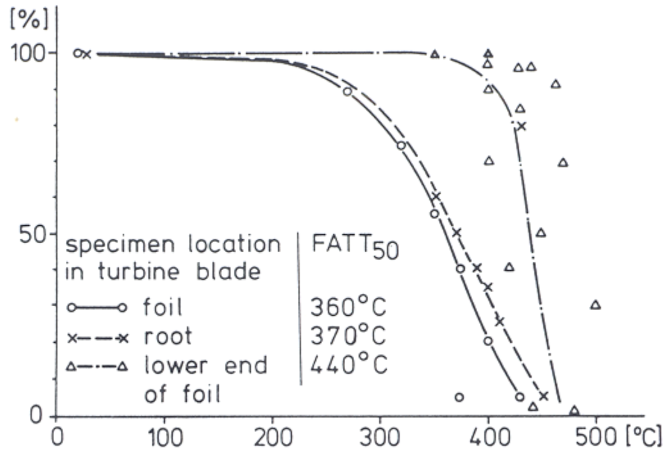

Figure 6 : Fraction of brittle surface of Charpy Impact Specimens versus test temperature for Mo-TZM alloy [Schuster 76].

5. EFFECT OF HELIUM ENVIRONMENT.

The main corrosion effects of metallic materials in HTGR helium environment have been extensively studied, especially for high temperature materials. Corrosion derives from the presence of low partial pressures of H_2O, CO, and CH_4 in helium. Hydrogen also plays a role in controlling the oxygen potential through the H_2/H_2O ratio. A recent review has given the following general trends [Graham 90].

- Below 850K (577°C), corrosion is limited. Only mild steels can suffer decarburisation. Steels with additions of chromium are slightly oxidised.
- From 850 to 1173K (577-900°C), creep resistant alloys form a chromium rich oxide scale. Internal microstructural changes are observed (precipitate free zones, carburised zones), especially when the surface scale is porous (figure 7).
- Under 1173K, rapid carburisation or decarburisation can occur, the dominant mechanism being controlled by the environment chemistry.

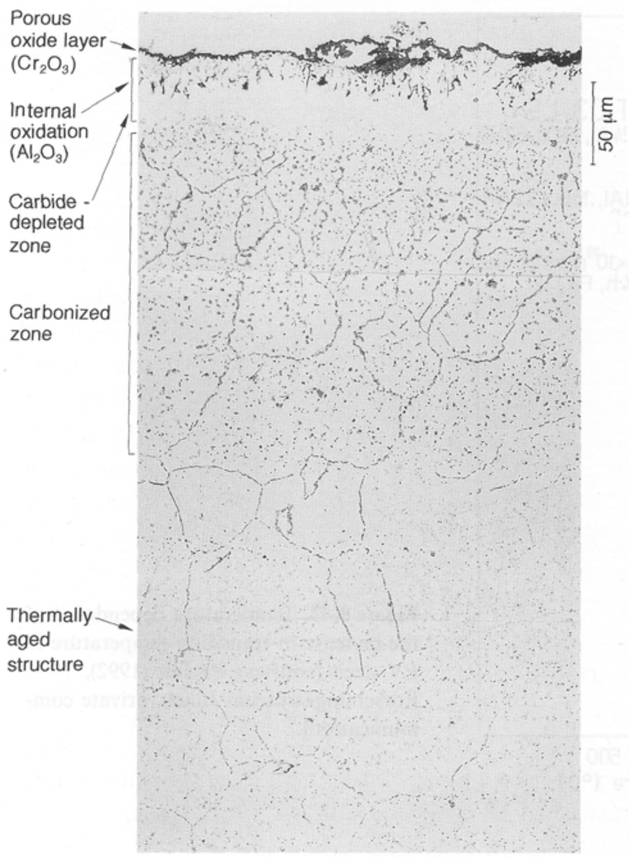

Figure 7 : Microstructure of alloy 617 after exposure for 1580h at 850°C to HTR helium [Dietz 94]

Concerning the high temperature alloys for blades and disks, the mains conclusions are:

- Mo-TZM alloy is almost insensitive to He environment up to 1000°C [Graham 76]. Jakobeit has observed a 30 µm thick MO_2C surface scale formed after 36000 hours at 850°C. This surface modification does not influence the creep resistance [Jakobeit 84];
- 713LC alloy and more generally nickel-base superalloys develop an oxide surface scale (Cr_2O_3) with an internal oxidation (Al_2O_3) and carburisation ($Cr_{23}C_6$). The carburised zone is depleted in γ' precipitates. Creep tests with 713LC specimens performed from 500 to 850°C in HTGR helium and air show virtually no environment effect on rupture time for test durations up to 55 000 hours [Jakobeit 84] (figure 8).

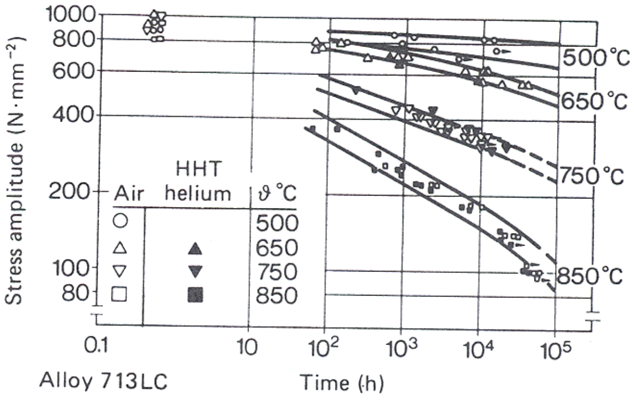

Figure 8 : Creep properties of IN713LC under air and HTR helium environments [Jakobeit 84].

The first way of optimisation to deal with corrosion problems is the optimisation of alloy composition, as shown in figure 9. Several chemical ratios were proposed as controlling parameters for the formation of stable protective surface scales.

- Alloys with high Al/Cr ratios (M21, IN591) are much more sensitive to corrosion than alloys with the same Al content and a reduced Cr content (Mar-M004, 713LC). Alloys with low Cr and high Ti develop a stable oxide scale. For this reason, alloys IN100 and IN738 appear promising for HTGR applications if the constraints on cobalt contents are lowered [Graham 76].
- Alloys with a ratio Al/Ti around unity develop more stable surface oxides than alloys with Al/Ti<1 (as 713LC, M21) [Johnson 84].

The second way to prevent corrosion problems is to apply a coating on the surface. An industrial experience exists for aluminide coatings for alloy 713LC, which then offers a stable continuous Al_2O_3 film.

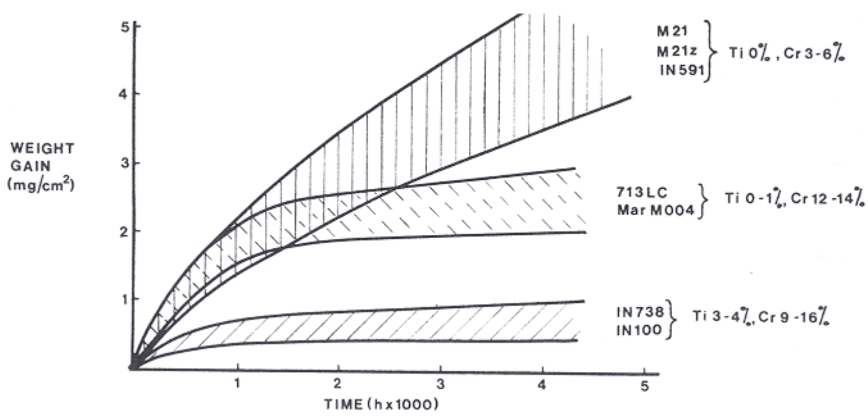

Figure 9 : Weight gain of cast Ni-base alloys in HTGR environment at 900°C [Graham76].

The main concern with design of high temperature alloys in impure helium is whether the surface corrosion influences the strength (and lifetime) of the components. From the literature available on creep rupture properties, it appears that degradation of mechanical properties after elevated exposure in helium is similar to degradation experienced in air [Jakobeit 84]. Therefore the degradation is mainly due to thermally activated structural changes rather than corrosion. However, when dealing with crack propagation properties, a study on cast Ni-base alloys has shown that environment effects are not similar in fatigue and creep crack propagation [Shahinian 86]. Complex mechanisms can occur at the crack tip (oxidation/plasticity interactions) which are difficult to model and predict.

6. LONG TERM MATERIALS PROPERTIES.

Most of the Ni-base superalloys investigated for turbine blades and disks are heat treated before use for maximum strength, usually with a solution treatment followed by aging treatments. They are strengthened by a very fine dispersion of coherent precipitates (γ’ or γ’’) with some carbides and borides. They are therefore inherently subjected to structural instabilities in high temperature service. Microstructural changes are of fundamental importance for HTGR material selection as most of the foreseen grades were developed for aeronautical turbines with much shorter lifetimes. For example, precipitation of hard brittle phases must be controlled for component lifetimes greater than 50000 hours.

Figure 10 shows an example of large microstructural changes of 713LC alloy after a creep test at high temperature. For the HTGR turbine, blade and disks alloys will endure thermal aging and there will be a strong need to understand the induced changes in strength, creep ductility and remaining life [Monma 89].

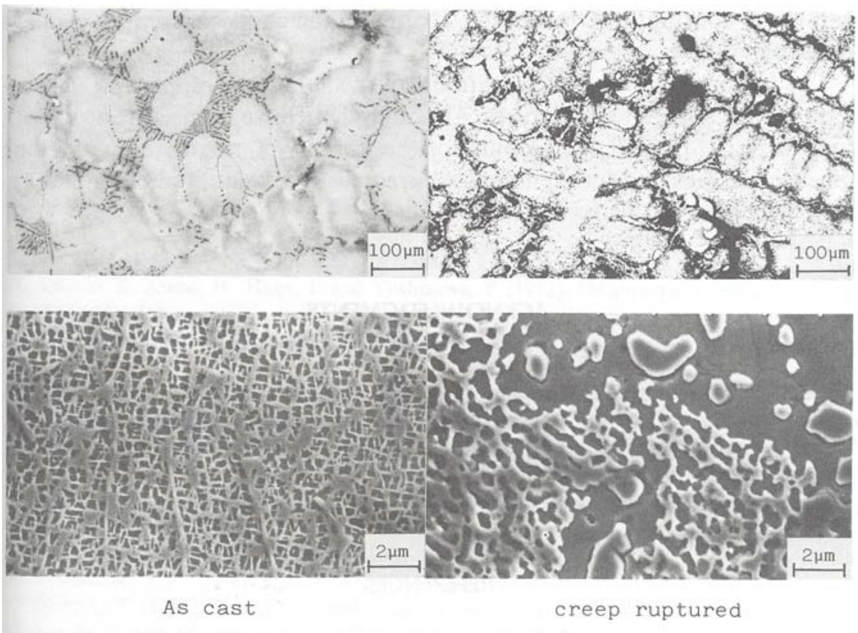

Figure 10 : Optical (upper) and scanning electron (lower) micrographs of alloy 713LC before (left) and after (right) a creep test at 1000°C for ~50000 hours [Monma 89].

7. POTENTIAL BENEFIT OF RECENT ADVANCES IN GAS TURBINE MATERIALS.

7.1 Evolutions in turbine technology.

There has been considerable development in natural gas fired advanced land-based turbines with combined cycles (reaching 60% thermal efficiency). These units are intended for base-load operations, with power continuously increasing : from 60-70 MW in the 70's to 280 MW today and around 500 MW in the next years (GE MS 7001H turbine in 2001). Reviews of recent achievements in turbine technology and associated materials developments can be found [Viswanathan 98, Materials 98].

The need to increase engine combustion temperature to increase the efficiency of gas turbines has resulted in a large increase in Turbine Inlet Temperature (TIT), for both land-based and aero turbines. In the last 50 years, the TIT has risen from around 800°C to nearly 1600°C [Viswanathan 98,Winstone 00]. This increase has been accompanied by several changes in the materials for the hottest parts. Recent advances in material processing and strength capability are now presented, as they certainly offer new possibilities to achieve the HTGR requirements.

7.2. Advances in disk materials.

This increase in TIT has a significant impact on turbine disks technology. Turbine disks rim temperatures have risen to 680°C and will exceed 700°C in future aero-engines. In the past, Waspaloy has been used extensively, and Alloy 718 and Udimet 720 are now replacing it. However, even U720 approaches its limits. Therefore, major efforts are made to produce disks with temperature capabilities above 700°C [Winstone 00].

Alloy 718 has a large propensity for segregation during casting. For aero-engines, melting technology advances have allowed the production of the required disk diameters of the largest engines. However, for power generation, turbine disks can be typically 1.5 meters or more in diameter. Such large components require the production of around 10 ton ingots with subsequent forging. This has not been possible with 718 alloy because of low castability, and a new grade with less Mo and Nb was developed (IN706). This grade allows the production of large disks with modest diminution in mechanical properties. However, melting technology of large 706 ingots remains complex, and the alloy reaches today the limits of its performance. Again, large efforts are put in place to produce large disks in more resistant 718 grade, and also Udimet 720 for the future [Oakes 00].

Large segregation during solidification of highly alloyed grades has forced a move to powder metallurgy processing for which segregation is not an issue. Alloys such as Merl 76 (Pratt &Whittney), René 88DT (General Electric) and N18 (Snecma) represent the current state of the art for aero-engines [Thomas 00].

In the past, HIPped powder compacts yielded excellent static properties, however low-life fatigue failure occurred in service. These failures were initiated at defects on powder particle boundaries due to powder contamination. To overcome these problems, atomisation and handling of the powders were improved, and post-HIP isothermal forging used to mitigate the influence of contamination. Today, powder metallurgy offers two interesting perspectives [Oakes 00].

- Production of Ni-base superalloy grades that are almost not castable (due to excessive segregation) and alloys with a low forgeability.
- Production of net shape parts with minimum final machining.

Recently, there has also been interest in the development of HTGR ceramic turbine disks in Japan [IAEA 00]. Achieving a rotor of lightweight and high strength could indeed facilitate the turbine design. With the HTGR helium environment, C/C composite was chosen, and the fabrication of a representative disk was achieved in 1998. Tests showed poor rotating properties, but further fabrication and tests are under way. Many technical problems still need to be resolved for this long-term solution of disk fabrication.

7.3. Advances in blade materials.

The rise in TIT has been met by replacing the forged blades by cast blades ($T^{100000h}_{140MPa} \approx 805°C$ and $\sigma^{850°C}_{100000h} \approx 175MPa$ for IN738 [Viswanathan 98, Thomas 00]), and the subsequent introduction of directionally solidified (DS) and single crystal (SC) blades ($T^{100000h}_{140MPa} \approx 870°C$ [18]). Today peak metal temperatures of over 1100°C are experienced in some turbine parts, with service lives around 10 000 hours achievable [Winstone 00].

The DS process allows blades to operate at temperatures approximately 25K higher than conventional blades. An additional temperature increase of approximately 25K is achieved by using blades made of single crystal materials (no grain boundaries). Thermal barrier coatings are extensively used in aircraft turbines, allowing another temperature increase of approximately 100K [Becker 00].

APPENDIX 1 : CHEMICAL COMPOSITION OF INVESTIGATED ALLOYS.

Possible alloys for blades (weight %).

Alloy	process	Ni	Cr	Co	Mo	Al	Ti	W	Nb	others
713LC	C	b	12.0	0.04	4.65	5.82	0.74	-	2.0	0.07C, 0.1 Zr, 0.009B
MAR-M 004	C	b	11.86	-	4.42	5.95	-	-	1.6	1.3Hf, 0.065C, 0.3Ta
IN 738	C	b	16	8.5	1.75	3.4	3.4	2.6	0.9	1.75Ta, 0.11C, 0.04Zr, 0.01B
IN 792	C	b	12.15	9.25	1.8	3.45	4.10	3.78	-	3.75Ta, 0.52Hf, 0.11C, 0.012B, 0.09Zr
René 80	C	b	13.8	9.4	3.9	3.15	5	3.8	-	0.16C, 0.017B, 0.04Zr
M21	C	b	6	-	2	6	-	10.5	1.5	0.1C, 0.02B, 0.1Zr
M21Z	C	b	6	-	2	6	-	10.5	1.5	0.1C, 0.02B, 0.5Zr
IN 591	C	b	3	12	-	5.7	-	19	-	0.1C, 0.03B, 0.37Zr, 3Ta
IN 100	C	b	10	15	3	4.7	3	-	-	0.18C, 0.014B, 0.06Zr, 1V
Udimet 700	C	b	15	18.5	5.2	4.3	3.5	-	-	0.08C
MAR-M 247	C, DS	b	8.5	-	-	5.5	1	10	-	0.15C, 3.0Ta, 1.3Hf
Nx-188	C	b	-	-	18	8	-	-	-	0.04C
Nimonic 80A	F	b	20	0.17	-	1.32	2.45	-	-	0.04C, 1.86Fe, 0.045Zr
Nimonic 90	F	b	19.5	16.5	-	1.4	2.35	-	-	0.09C
Nimonic 105	F	b	15.4	20.25	4.72	4.69	1.00	-	-	0.14C, 0.051Zr
FIS 145	F	b	7.44	9.90	-	5.28	-	13.46	-	0.11C, 0.88Hf
Nimonic 115	F	b	14.2	13.2	3.2	4.8	3.7	-	-	0.16C
Udimet 520	F	b	19	12	6	2	3	1	-	0.05C, 0.005B
René N4	C, SC	b	9	8	2	3.7	4.2	6	0.5	4Ta
Mo-TZM	C, P	-	-	-	b	-	0.45	-	-	0.015C, 0.09Zr

F : wrought or forged, C : as Cast, P : Powder Metallurgy, DS : Directionally Solidified, SC : Single Crystal.

Possible alloys for disks (weight %).

Alloy	process	Ni	Cr	Co	Mo	Al	Ti	W	Nb	others
A 286	F	26	15	-	1.3	0.2	2.0	-		Fe base, 0.05C, 0.015B
IN 706	F	41.5	16	-	0.5	0.2	1.0	-	2.9	Fe base, 0.03C,
Waspaloy	F	b	19.5	13.5	4.3	1.3	3.0	-	-	0.08C, 0.006B, 0.06Zr
Udimet 720 (PER 72)	F, P	b	17.9	14.7	3.0	2.5	5	1.3	-	0.03C, 0.033B, 0.03Zr
Merl 76	P	b	12	18.5	3	5	4	-	1.5	0.02C, 0.5Hf
René 88DT	P	b	16	12.7	4	2.15	3.7	4	0.7	0.05C, 0.05Zr
N18	P	b	11.2	15.6	6.5	4.4	4.4	-		0.13Fe, 0.03Zr, 0.5Hf, 0.015B, 0.018C
IN 718	F	b	19	-	3	0.6	0.8		5.2	19Fe
MA6000	P	b	15	-	2.0	4.5	2.5	4.0	-	2.0Ta, 0.05C, 0.01B, 0.15Zr, 1.1Y2O3

F : Forged, C : as Cast, P : Powder Metallurgy.

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