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HTR-M1 Turbine Materials - Summary Report

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HTR-M1 Turbine Materials - Summary Report

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CONTENTS

FIGURES

TABLES

SUMMARY

This document contributes to the 5th Framework Project of HTR-M1 and summarises the results of the HTR-M1 investigations on turbine materials (Deliverable D6).

The work includes an extension of the activities started within the HTR-M Project covering further testing on creep and on treated turbine blade and disc materials to assess the effects of corrosion (carburisation/ decarburisation) covering short and intermediate times plus aging tests.

The tests carried out in air endorse the properties of the selected turbine blade (IN 792 DS, CM 247 LC DS) and disc (Udimet 720) materials and cover heat treated, aged, carburised and de-carburised conditions to check the effects on deterioration of strength properties.

The results suggest that CM 247 DS is the better material in the as-received and treated conditions. Further test are recommended however for longer times to substantiate the properties and check the suitability for 60,000h design life.

The long-term structural stability of the materials needs to be further evaluated. Thermal ageing of approximately one years duration had a detrimental effect on properties and longer term aging tests are recommended to confirm suitability for long service times.

Carburisation had the most noticeable effect on properties giving a 30% reduction in yield strength and a marked reduction in creep rupture and fatigue life.

Additional tests are recommended for the de-carburised condition to reinforce the apparent conclusion that this has little effect on material creep strength.

Further testing is continuing on the remaining available specimens. Additional tests and further micro-structural evaluation is needed to re-enforce the findings from these few available tests.

1.0 INTRODUCTION

This document contributes to the 5th Framework Project HTR-M1, Work Package 1 on turbine materials and summarises the results of the technical and test work performed. The activities extend the materials platform established in HTR-M [Ref. 1] by considering the effects of creep damage and environment focussing on intermediate creep duration testing. The activities cover the following areas of work:

- Results and assessment of tests performed at CEA on materials identified in HTR-M for the turbine blade and disc materials at remaining short and at medium times plus ageing tests
- Results and assessment of tests performed at JRC on the above treated turbine blade and disc materials to asses the effects of corrosion (carburisation/ decarburisation)

The Project objectives with respect to the investigation of turbine disc and blade materials are summarised together with overall conclusions and summary of any remaining items to be investigated within the Raphael-IP 6th Framework Project.

This document contributes to the 5th Framework Project of HTR-M1 and provides a summary of the results obtain on Turbine materials (Deliverable D5 – Table 0-1).

2.0 HTR-M1 TURBINE MATERIAL WORK OBJECTIVES

Reliable material property data is a key requirement in the development of HTR and VHTR technology. An understanding of material behaviour plus materials data obtained under representative reactor operating conditions is required for the main HTR components in order to assess safety and feasibility. Work Package 1 of the HTR-M1 project deals with the selection and development of materials data for the most highly loaded areas of the turbine (direct cycle). The following provides an overview of the HTR-M1 tests on turbine disc and blade materials:

2.1 Work package 1 of HTR-M1 – Intermediate creep tests on turbine materials

Turbine materials can operate in helium at temperatures of 850-900°C (HTR) and up to 1000°C (VHTR). High strength alloys are being considered for the blades that are capable of withstanding such temperatures and conditions in the direct cycle environment for operating periods approaching 60,000h. Creep strength and the corrosive effects of helium impurities on the properties and strength of the materials are important issues that must be evaluated. Manufacturing considerations are especially important and candidate alloys must have good properties and proven thermal stability. For the blades, cast material is not sufficient, and so directionally solidified alloys have been considered. The HTR-M investigations have highlighted the need for cooling and possibly coatings for both disc and blade alloys. Impurities in the helium at temperature can result in carburisation and/or de-carburisation of the super-alloys and lead to a lessening of material strength. From HTR-M both an Aluminium former and Chromium former were selected as potential options for the 850^{o}C gas temperature case (Section 3).

The turbine test programme within HTR-M1 covers medium term creep tests under simulated HTR environments. The impact of corrosion is an important selection criterion since the coolant gas used in an HTR contains small levels of impurities (H_2, H_2O, CO, CO_2) at low partial pressures, and these can interact with the core graphite and metallic components and cause degradation of their properties. The blade materials chosen were CM 247 LC DS and IN 792 DS with tests conducted using four simulated environments (see Section 3):

Within HTR-M1 the blade tests are to be performed for periods of up to 10,000h and at temperatures up to 850^{o}C. HTR-M results in air show material strength data to be consistent with available published results and the CM 247 LC DS to have the better high temperature properties for the periods tested. Tests on treated materials remain to be completed to confirm this. All treatments and mechanical testing will be completed as far as possible within the extension period of the HTR-M1 project (see below).

3.0 SELECTION, PROCUREMENT AND TEST PROGRAMME

This section reviews the tasks performed for the intermediate creep testing of turbine blade materials under simulated reactor conditions. A description of the test programme and planning for the tests is given in HTR-M1 02/9-D-1-1-7 (Deliverable 1). A report on specimen fabrication, procurement and testing requirements (see Table 0-2) is given in HTR-M1 03/3-D-1-0-12 (See also Deliverable D2 of Table 0-1)

The HTR-M1 creep tests are carried out on two turbine blade alloys (CM 247 LC DS and IN 792 DS) for durations of up to 10,000h at 850^{o}C. Note that tensile and short-term creep tests on these alloys at temperatures of 850^{o}C and for creep periods of up to 300h are available from the HTR-M project. They provide a basis for the estimated stresses and effect of environment (Figures 4-1 to 4-3).

The Directionally Solidified Blade materials were supplied by Doncasters GmbH and the specimens were machined and heat treated (16 mm dia. X 150 mm length):

- 48 bars of IN 792 DS (heat RC415) received in March 2003 with a solution heat treatment (vacuum 1180°C/2h, cooling rate 30K/minute).

- 30 bars of CM 247 LC DS (heat X4012) received in may 2003 with a solution heat treatment and aging (vacuum 1220°C/2h + 1240°C/2h + 1246°C/2h cooling rate 80K/minute + vacuum 1080°C/4h/forced gas circulation + 900°C/8h/ forced gas circulation).

All test bars were subjected to a Dye Penetrant Inspection Examination and X-Ray examination to ensure that no cracks or open porosities were present.
The chemical composition of the test bars is given below in Table 3-1.

Table 3-1 Chemical Composition of IN 792 DS and CM 247 LC DS Blade Materials. (Weight %).

	C	Si	Mn	Cr	Mo	Ni	Ta	Ti	W	Co	Fe	Al	Hf	V	Zr
CM 247	0.074	<0.03	<0.03	8.16	0.43	Bal.	3.21	0.68	9.55	9.364	0.03	5.63	1.42	<0.03	0.015
	Cu	**B**	**P**	**S**	**Pb**	**Ag**	**Bi**	**Te**	**Tl**	**Sb**	**Sn**	**Zn**	**Cd**		
	<0.03	0.0155	<0.005	6 ppm	<2ppm	<2ppm	<0.3ppm	<1ppm	<1ppm	<3ppm	<3ppm	<2ppm	<0.5ppm		

	C	Si	Mn	Cr	Mo	Ni	Nb	Ti	Ta	Al	Fe	Co	Cu	Hf	W
IN 792	0.079	<0.05	<0.05	12.63	1.82	Bal.	<0.1	3.91	4.18	3.52	0.02	9.08	0.03	1.34	4.25
	B	**Zr**	**P**	**S**	**Ag**	**As**	**Bi**	**Mg**	**N**	**O**	**Pb**	**Se**	**Te**	**Tl**	
	0.017	0.02	<0.002	<20ppm	<0.1ppm	<5ppm	<0.1ppm	<6ppm	<11ppm	<12ppm	<0.3ppm	<2ppm	<1ppm	<0.2ppm	

Initially, it was proposed to carry out mechanical tests (creep tests on notched sample) plus micro-structural investigations (crack lengths, orientations, location...) to study damage evolution in the blade alloys and identify parameters required for damage models. However, following discussions (held principally during a specialist meeting prior to the 3rd HTR-M1 & M1 progress meeting (Refs. 1 & 2)) it was agreed to abandon these since with this programme only two tests per environment for each grade (air and HTR helium containing impurities) would be possible. At this meeting it was also agreed, since the helium atmosphere in the reactor is largely unknown and the results may be sensitive to atmosphere changes, to abandon the "representative helium" condition and instead to test the specimens in four states as indicated below.

- As received with ageing heat treatment – I
- Decarburised – II
- Carburised – III
- After the carburisation or decarburisation heat treatment – I(mod)

It was agreed that this approach would be applied to both the HTR-M and HTR-M1 tests on both the turbine disc and blade materials.

Originally it was proposed to carry out the tests for conditions I and I(mod) at CEA in air, and conditions II and III at JRC in argon. After some discussion it was decided that all tests would be carried out in air. Ageing would be carried out in an argon atmosphere, using oversized blanks.

The test matrices and proposals for the test conditions were as far as possible planned so that they bound the temperature and transient cases considered for the first row of the turbine but using a maximum temperature of 850^{o}C.

The proposed test matrix for the HTR-M1 turbine material tests is given in Table 0-3. Damage analysis is to be confined to the following activities:

- observation of fracture surface by SEM;
- identification of damage type (intergranular, transgranular) on longitudinal cuts;
- identification of damage parameters that would be suitable to fit the creep data using the usual creep damage models for super-alloys (i.e. MacLean and Dyson models).

The start of the medium term creep tests on the treated specimens was planned from early 2004 following the completion of the short-term HTR-M tests. However problems during the construction and commissioning of the corrosion rig at JRC (Figure 5-1) resulted in significant delays to the treating of the specimens for conditions (II) and (III) and only the tests in air at CEA could be started to programme. The testing of specimens in the de-carburised (II), carburised (III) conditions had to be put back until the corrosion rig was fully operational.

The conclusion from the short term HTR-M results in air [Ref. 4] was that CM 247 LC DS has the better high temperature properties (than IN 792 DS) and therefore may be the best candidate. This remains to be confirmed however by the follow-on tests on the treated specimens and the HTR-M1 results.

An extension to the HTR-M1 Programme was obtained so that the results of the graphite irradiation tests could be fully reported. This also enabled any further creep tests on blade alloys started within HTR-M (long term tests + condition denoted as Imod in the test matrix) and the planned HTR-M1 tests (creep and tensile properties after long term ageing for IN 792) within the HTR-M1 extension period. The schedule was revised to complete as much of the treatments and mechanical testing as possible.

<u>Heat treatment problem with IN 792 DS :</u>

It was discovered in September 2003 that the IN 792 DS material was not properly heat treated by the manufacturer (a second solution step (1080^{o}C/4h) and the final ageing (850^{o}C/ 24h) was missing. At the time some IN 792 bars had already been machined at CEA and some others had started a 10000h thermal ageing at 850^{o}C within the HTR-M1 Programme. It was therefore decided to restart the whole heat treatment on the aged bars and to simply add the missing heat treatment steps to the other bars.

4.0 EXPERIMENTAL RESULTS ON BLADE MATERIALS

4.1 Metallography

Metallographic and TEM examinations were performed on IN 792 DSand CM 247 LC DS, which were subsequently reported in HTR-M deliverable 2.10 [Ref. 4]. The main conclusions from the micro-structural examination of the blade alloys were as follows:

The structure of CM 247 LC DS was observed by optical microscopy for sections parallel and perpendicular to the grains. The grain average width was 150 µm. For the structure parallel to the grains, misorientations were present also carbides and areas with primary eutectic γ' phase. For the normal orientation residues of casting micro segregation and dendrite structure, and a few micro--porosities were present. The examination was typical of DS blade alloys, CM247 LC alloy micro-structural aspects.

For IN 792 DS there were no microporosities in sections parallel and perpendicular to the grains , however chemical segregation was observed in the transverse direction of a more pronounced nature than for CM 247 LC DS.

TEM observations on CM 247 LC DS showed gamma prime precipitates in the gamma matrix. The γ' size was uniform except in areas close to the large primary γ' where a very small γ' distribution was seen. Low angle grain boundaries are decorated with small precipitates that have been identified as phases enriched in W, Cr and Ni. These precipitates seem to have a chemical composition corresponding to the so-called "µ phase". A few very large precipitates are also dispersed in the matrix. These precipitates were identified as carbides enriched in Hf and Ta.

4.2 Tensile properties in Air

Tensile tests were performed in air at 20, 750, 800 and 850°C with a strain rate of 5.10^{-4} s^{-1} for both the CM 247 LC DS and IN 792 DS materials. These tests would also provide results for comparison with the tests in the metallurgical states II, III and I_{mod} (Section 3).

Figure 4-1 presents results for the yield stress ($R_{0.2\%}$) for both grades at various temperatures. CM 247 LC DS shows a better yield strength than IN 792 DS, especially at 850°C. The same trend is seen in Figure 4-2 for Ultimate Tensile Strength (UTS).

For the rupture strain, the two grades show equivalent data except at room temperature where the IN 792 DS ductility is rather low (Figure4-3).

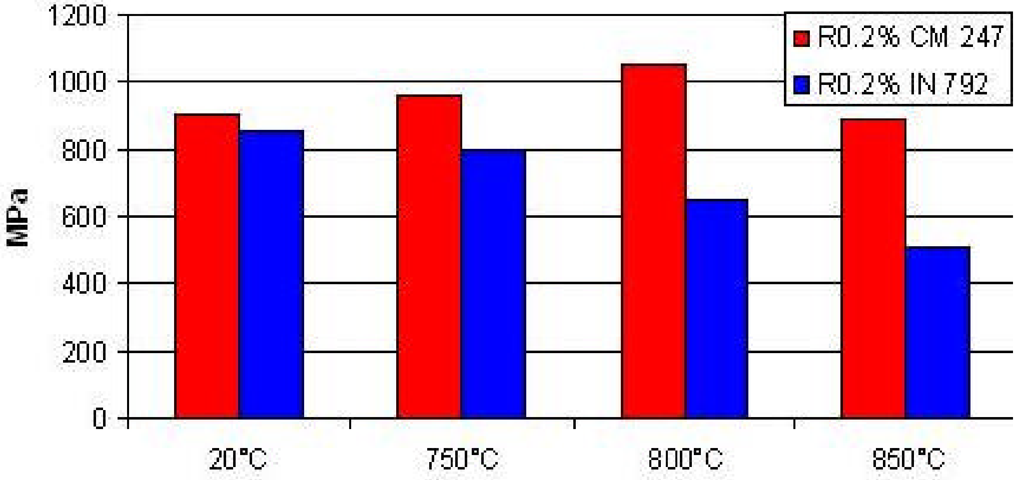


Figure 4-1 Comparison of 0.2% Yield Stress for CM 247 LC DS and IN 792 DS from 20°C to 850°C.

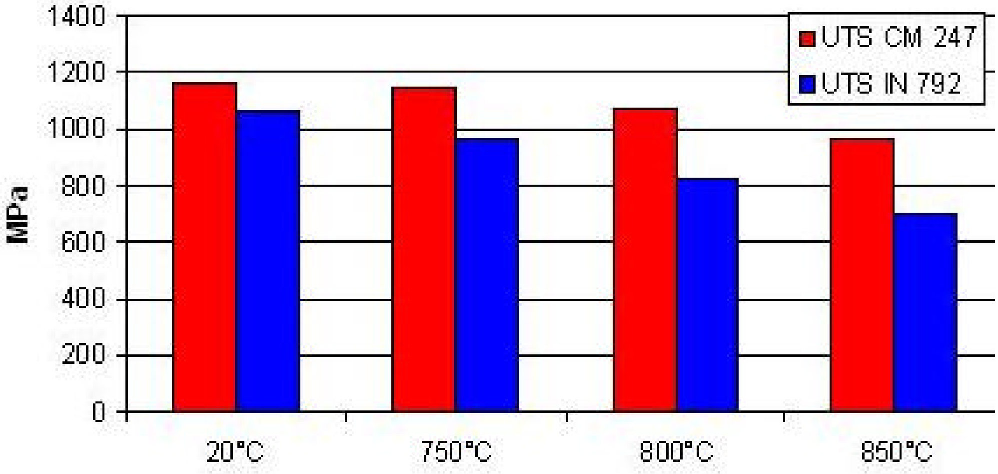


Figure 4-2 Comparison of Ultimate Tensile Stress for CM 247 LC DS and IN 792 DS from 20°C to 850°C.

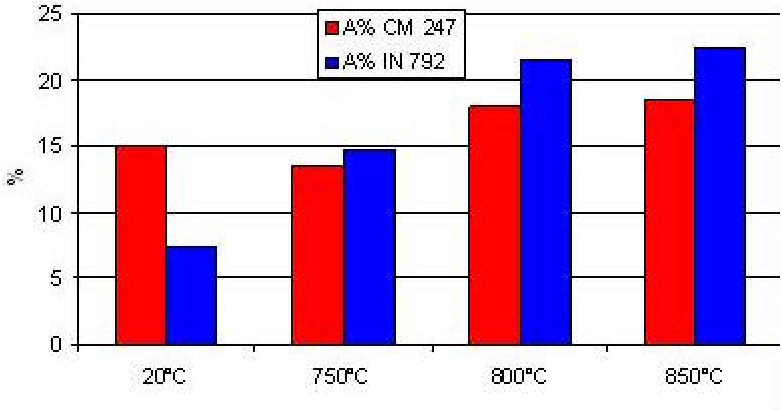


Figure 4-3 Comparison of elongation at rupture for CM 247 LC DS and IN 792 DS from 20°C to 850°C.

4.3 Creep properties in Air

As for the tensile tests, these results serve as a basis for comparison with the tests in the metallurgical states II, III and I_{mod}.

All the long term creep tests started within HTR-M have ended by the failure of the specimens, except for one test on CM 247 LC DS which had to be interrupted before failure in order to start the HTR-M1 tests on the corresponding creep machine.

Creep tests were performed on both IN 792 DS and CM 247 LC DS grades at 850°C in air. Creep stresses were selected to match the rupture times planned in the test matrix (~1000h, ~3000h and ~9000h).

The stresses selected for these rupture times (1000h, 3000h and ~9000 h) at 850°C were based on bibliographic data found on these materials [Ref. 5]:

- CM 247 LC DS: 275 MPa (9000h), 325 MPa (3000h) and 375 MPa (1000 h)
- IN 792 DS: 225 MPa (9000h), 275 MPa (3000h) and 325 MPa (1000 h)

Figure 4-4 shows the creep curves obtained for CM 247 LC DS at 850°C. The targeted times to failures are reasonably well attained. For all tests, it was noticed that the elongation at rupture was larger than 20%. The corresponding creep curves for IN 792 DS are shown in Figure 4-5. Again as for CM 247 LC DS, the ductility is corrected for all tests

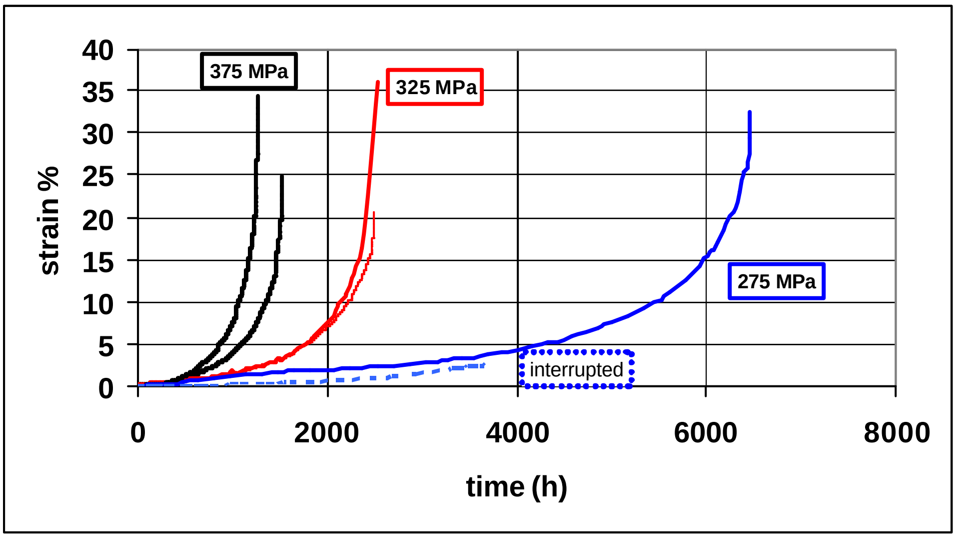


Figure 4-4 Creep curves obtained at 850°C for CM 247 LC DS grade.

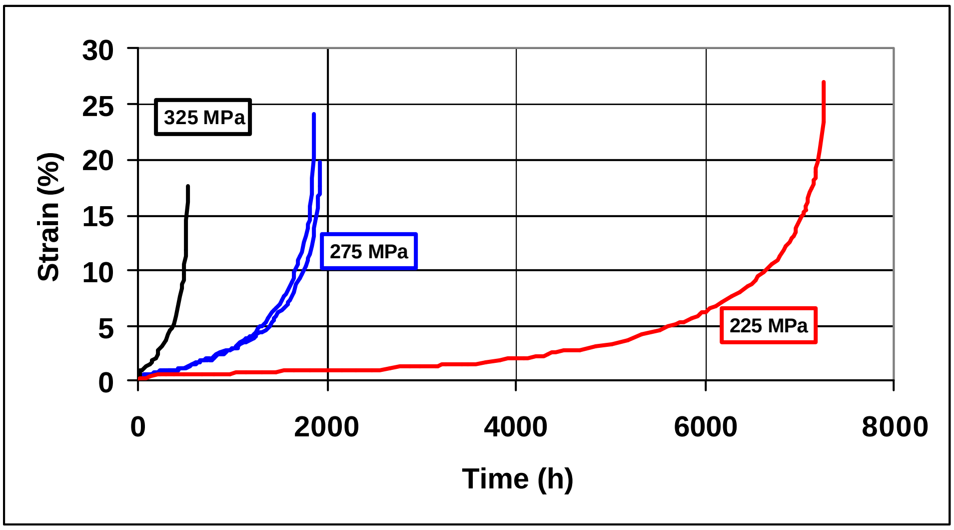


Figure 4-5 Creep curves obtained at 850°C for IN 792 DS grade.

Figure 4-6 below compares the measured creep rupture times with results from the bibliography [Ref. 5]. It shows that the HTR-M1 results are consistent with bibliography data for the two blade alloys tested. As for tensile properties, it was also noticed that the creep strength of CM 247 LC DS was superior to that of IN 792 DS in air.

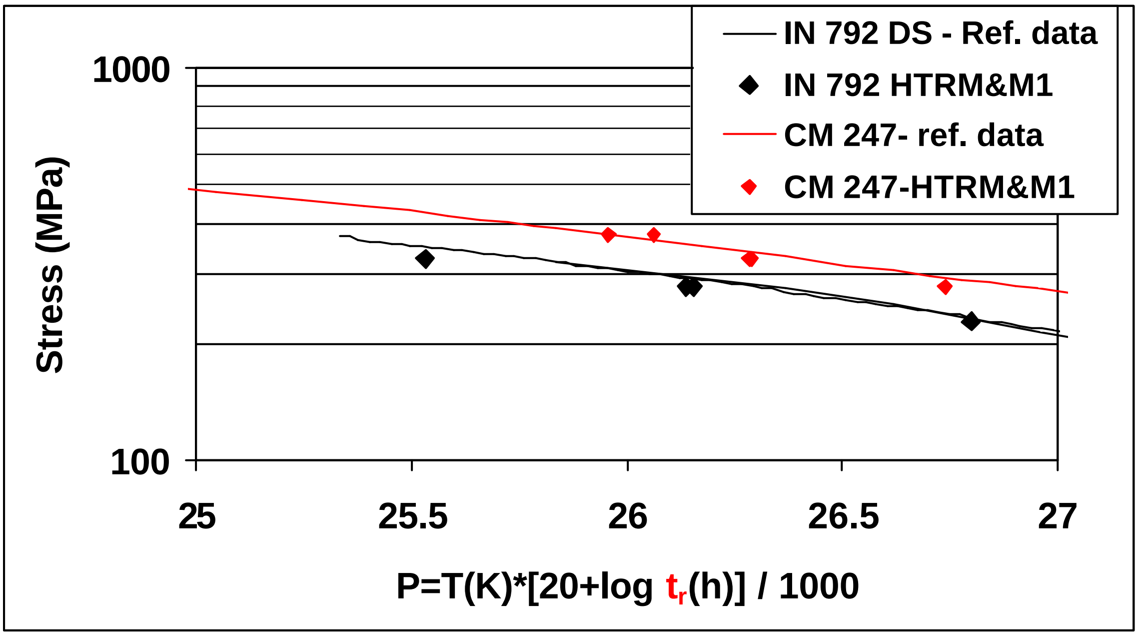
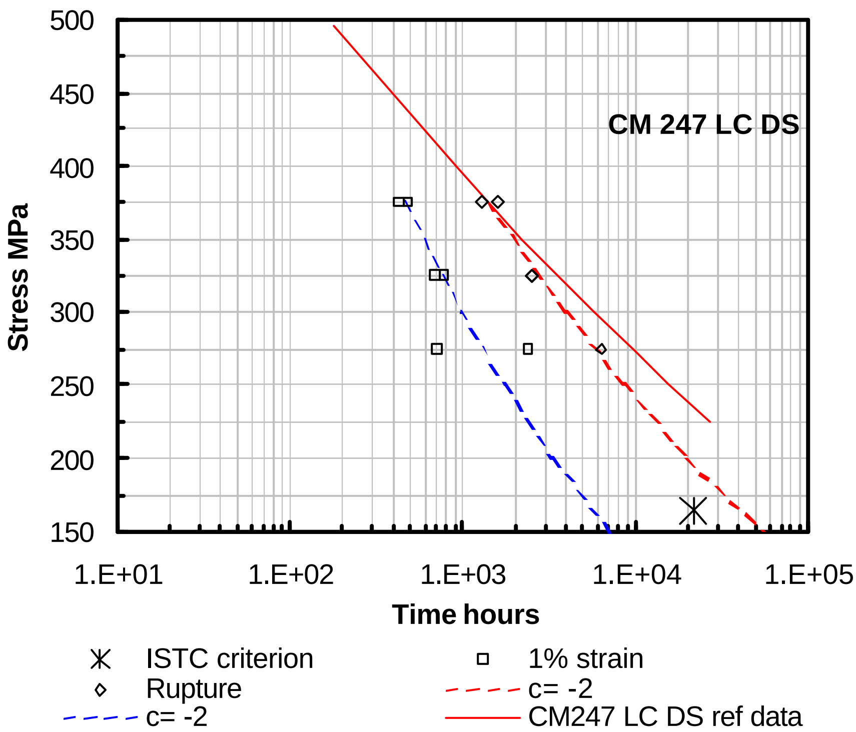


Figure 4-6 Comparison of CM 247 LC DS and IN 792 DS creep rupture properties, with the Larson-Miller parameter calculated at rupture.

The work of the Russian team within ISTC on the design of the GT-MHR turbine rotor [Ref. 6] suggested that the calculated minimum required properties for the first stage blades was 164 MPa at 830°C for 1% creep strain. Conversion to 850°C using the Larson-Miller parameter T(K)(20+log t(h)) gives a target time to 1% creep strain of 21700 hours, about an order of magnitude beyond the data obtained in this work. Comparisons with the ISTC target were therefore made by extrapolation of curves fitted to the 1% creep data based on the following equation:

$$\log_e t = A + B\sigma + C \log \sigma$$

In view of the scarcity of data a fixed value of c= -2 was chosen based on experience with other materials and the reference rupture data for the two blade alloys. The fitted curves are shown in Figure 4-7. They suggest that CM 247 LC DS may lose its strength advantage over IN 792 DS at longer times. Longer term tests are required to confirm this trend and to determine whether one or both alloys can meet the ISTC requirement with an acceptable margin.



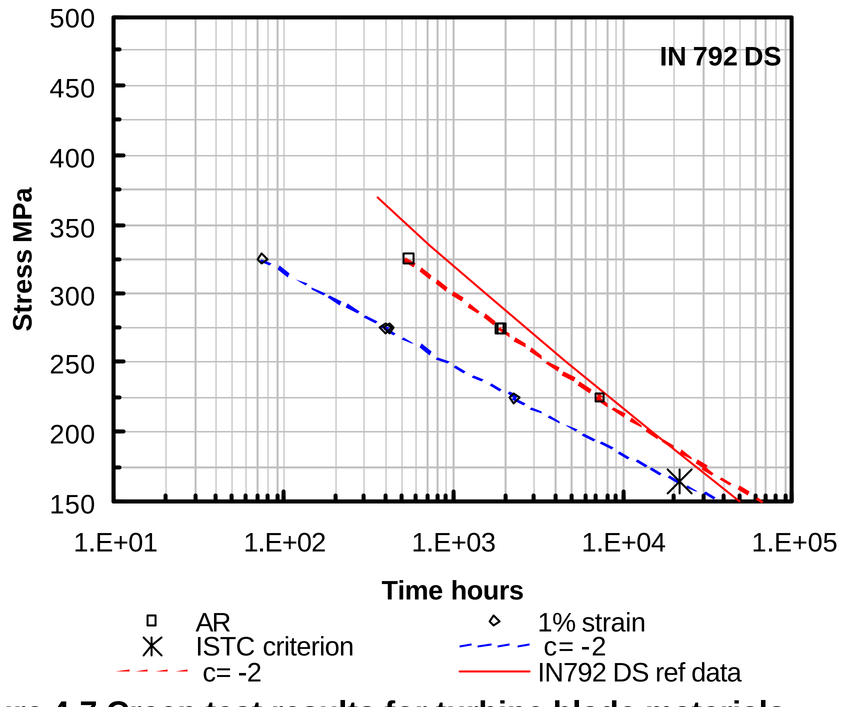


Figure 4-7 Creep test results for turbine blade materials

4.3.1 Fracture Analysis in Air

The fracture surface of a CM 247 LC DS creep specimen (longest achieved) tested at 850°C/275 MPa (tr = 6340 h) was examined using Electron Microscopy. The failure mode was ductile with numerous dimples shown on the fracture surface. Note that similar observations were also obtained for the IN 792 DS specimens. For the CM 247 LC DS material secondary cracks were observed mainly in areas were chemical segregation was present. Cracks close to the surface were oxidised, with the internal penetration rather limited, indicating that CM 247 LC DS has a good resistance to air oxidation (largely due to the formation of alumina-rich oxide when exposed to air). For an IN 792 DS specimen after a creep tests at 850°C/225 MPa (tr = 7250 h), again cracks were observed mainly in the interdendritic areas but here surface oxidation was more pronounced and a 50 μm thick band without γ' precipitates was observed beneath the surface oxide. IN 792 DS, a chromium oxide former, has a lower oxidation resistance in air.

4.4 Results on aged specimens in Air

The HTR-M1 mechanical tests on aged specimens were performed on IN 792 DS and CM 247 LC DS after two types of ageing treatments:

- Following the carburization and decarburisation heat treatment cycle defined by JRC (950°C/1000h) for both IN 792 DS and CM 247 LC DS (condition denoted I_{mod} in test matrix).

- After a thermal ageing of ~5000 h and ~8000 h at 850°C for IN 792 DS only.

The effect of these heat treatments on tensile and creep properties are summarised in the following sections:

4.4.1 Tensile properties after I_{mod} treatment in Argon

Tensile tests were conducted at 20, 750, 800 and 850°C using a strain rate of 5.10^{-4} s^{-1}. Figure 4-8 presents results for the 0.2% yield stress of CM 247 LC DS and IN 792 DS before and after the I_{mod} heat treatment. For CM 247 LC DS material, the I_{mod} heat treatment results in a decrease of the yield strength, from about -10% to -15% on the test temperature range. Figure 4-9 also shows that the ultimate tensile strength is slightly decreased (-5% to -10%). Note also that as shown in Figure 4-10 the effect on ductility is not significant.

For IN 792 DS, the effect of the I_{mod} heat treatment results in a slight decrease of yield strength and ultimate tensile strength at room temperature, and a slight increase of strength at high temperature. This effect could be explained as a result of the I_{mod} heat treatment partially solutioning the γ' precipitates. The microstructure of

IN 792 DS after the heat treatment is in this case strongly dependant on the cooling rate.

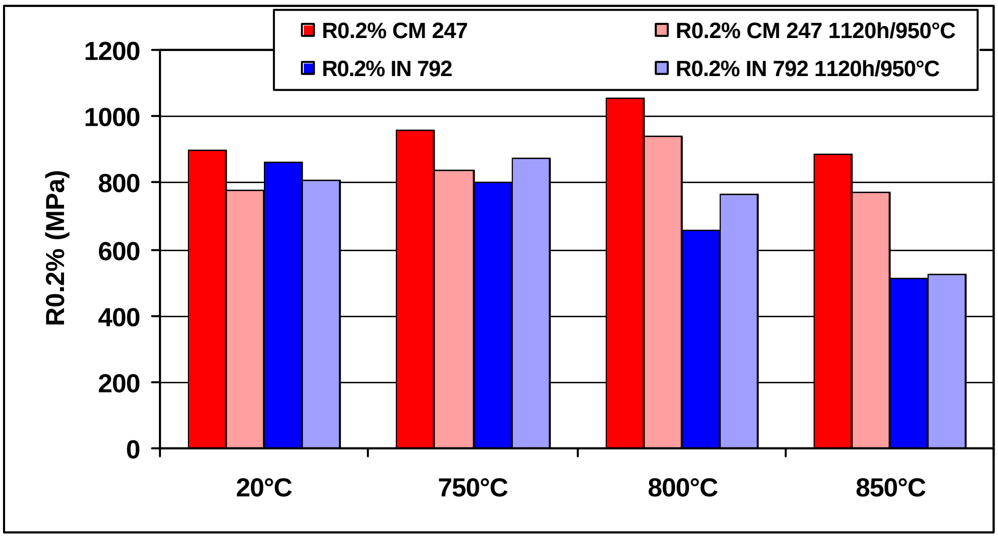


Figure 4-8 Comparison of 0.2% Yield Stress for CM 247 LC DS and IN 792 DS from 20°C to 850°C, in the as-received condition and after the I_{mod} heat treatment.

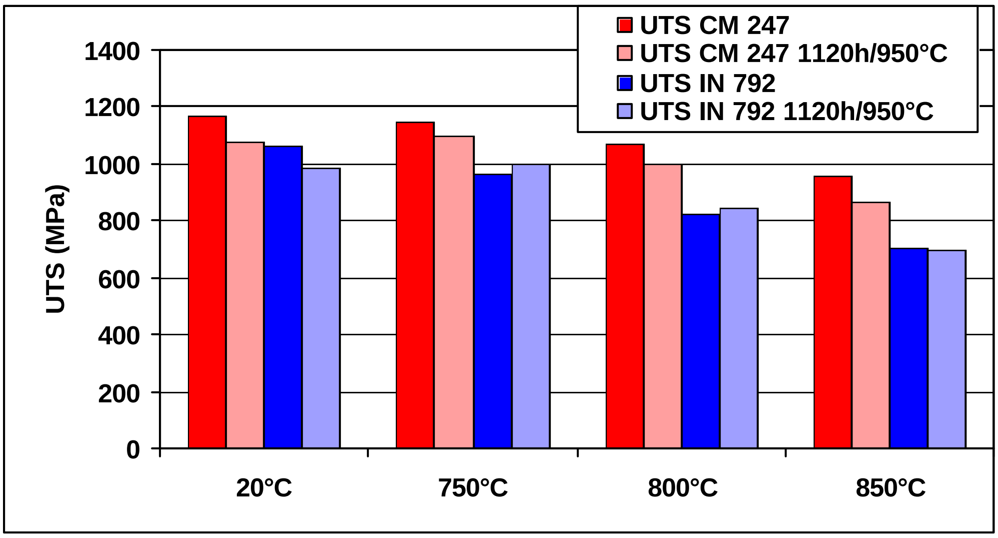


Figure 4-9 Comparison of Ultimate Tensile Strength for CM 247 LC DS and IN 792 DS from 20°C to 850°C, in the as-received condition and after the I_{mod} heat treatment.

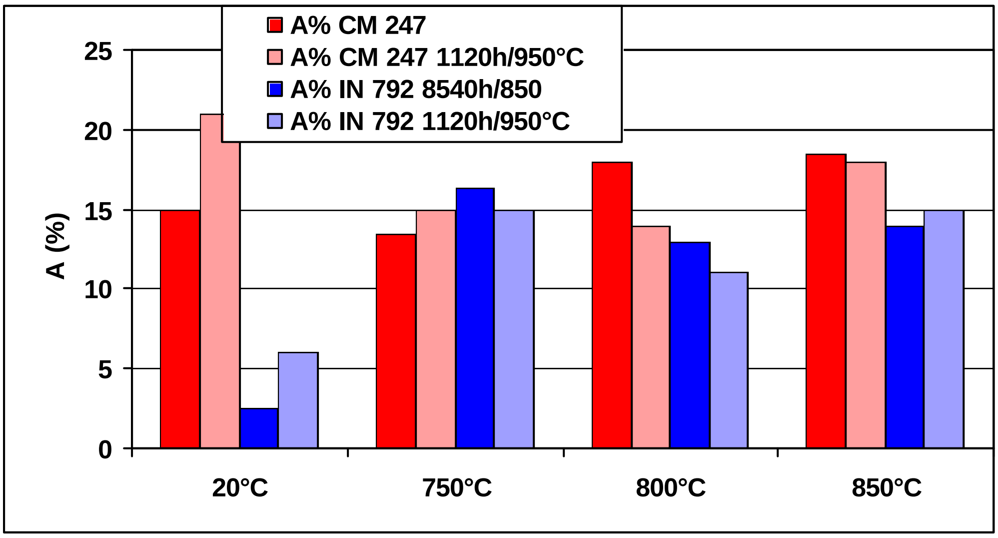


Figure 4-10 Comparison of elongation at rupture for CM 247 LC DS and IN 792 DS from 20°C to 850°C, in the as received condition and after the I_{mod} heat treatment.

4.4.2 Creep properties after I_{mod} treatment in Argon

The creep properties after the carburisation heat treatment have been assessed for the two grades, and the results are presented in Figure 4-11and Figure 4-12. For CM 247 LC DS material, the I_{mod} heat treatment has a strong effect on creep lifetime, with a reduction by a factor 2. For IN 792 DS, only two tests were performed after I_{mod} heat treatment, and the longest one had to be stopped at the end of HTR-M1 project. The test at 275 MPa also showed an effect of I_{mod} heat treatment on lifetime, but ductility is unchanged.

To conclude, it was shown that the I_{mod} heat treatment decreases creep life time but has little or no effect on creep ductility for the two blade alloy grades evaluated in HTR-M1.

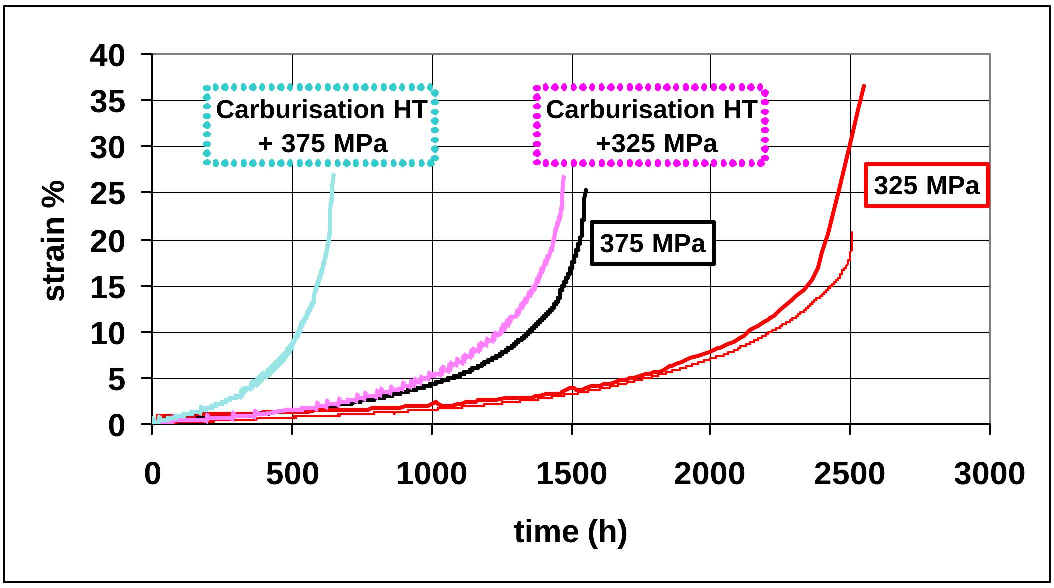


Figure 4-11 Creep curves obtained at 850°C MPa for CM 247 LC DS specimens with and without prior I_{mod} Heat Treatment.

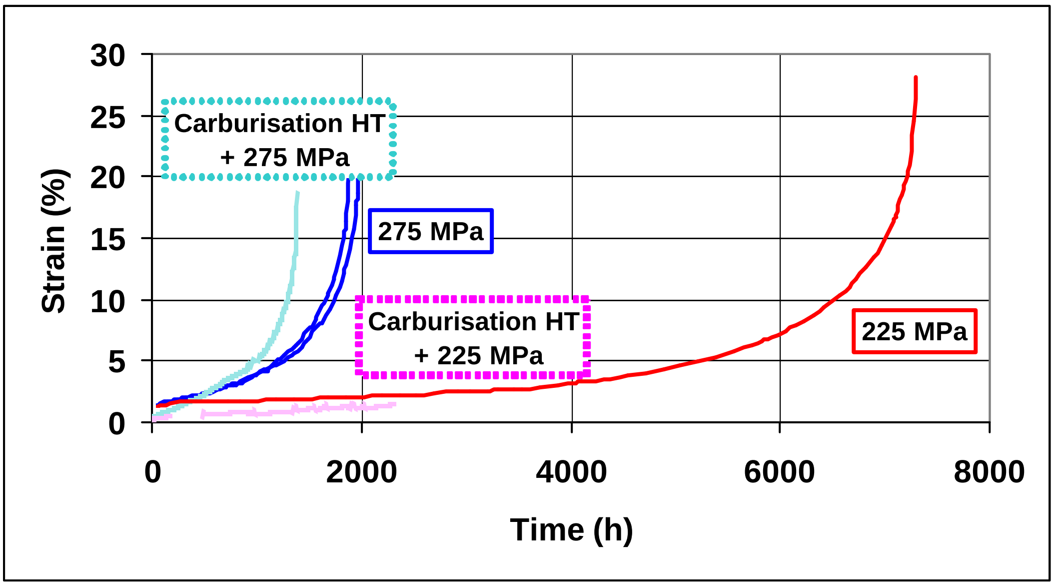


Figure 4-12 Creep curves obtained at 850°C MPa for IN 792 DS specimens with and without prior I_{mod} Heat Treatment.

4.4.3 Thermal ageing heat treatment (850°C) on tensile properties

The thermal ageing heat treatment was performed in air (on specimen blanks) for IN792 DS only, as planned in the test matrix. Two aging duration of 5000h and

8540h were finally attained within the programme time available in order to provide an evaluation of the ageing effects on the mechanical properties.

The effect of the two thermal aging durations on tensile properties are shown in Figures 4-13 to 4-15 which provide a comparison of results for values of Yield stress, UTS and ductility (see below).

The effect of a long term ageing at the service temperature (850°C) results in only a slight decrease in strength. Ductility is also decreased, except for the tensile tests performed at 750°C. Also the room temperature ductility of IN 792 DS after 850°C/8540 h aging is low, and this would certainly not be acceptable for a turbine within a nuclear plant application. An explanation for this is the slight coarsening of the γ' precipitates during the thermal ageing, together with the precipitation of few phases (carbides, TCP phases) having an embrittling effect.

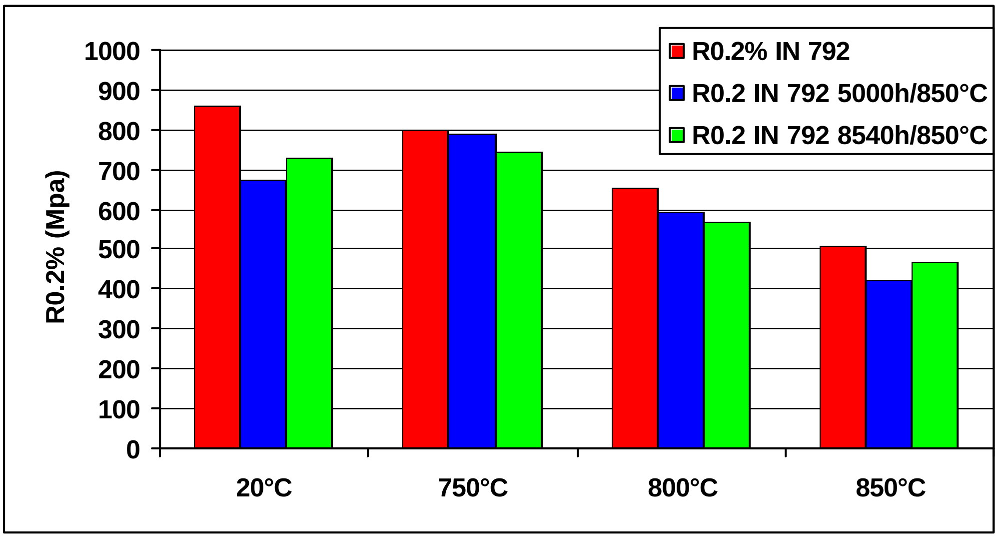


Figure 4-13 0.2% Yield Stress for IN 792 DS from 20°C to 850°C, in the as received condition and after 5000h and 8540 h of thermal aging at 850°C.

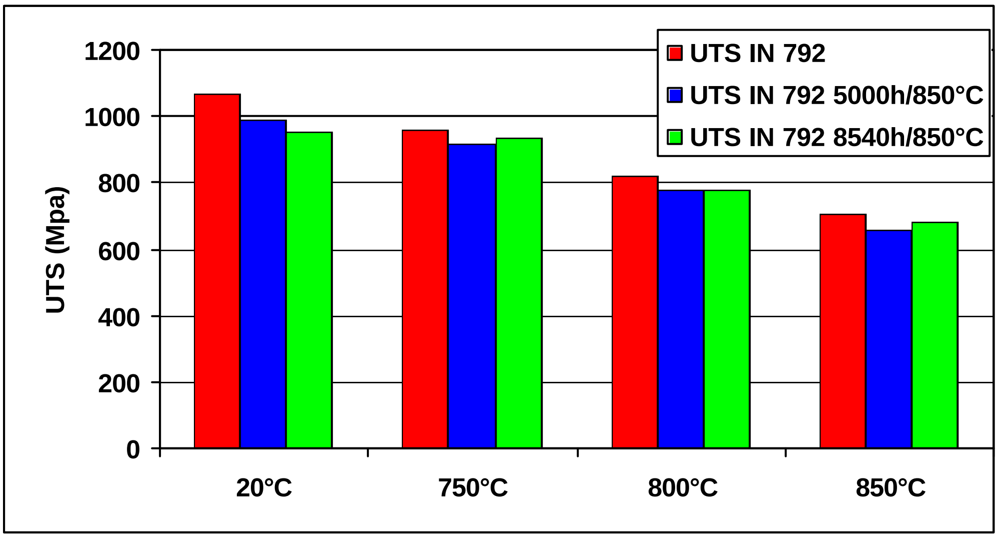


Figure 4-14 Ultimate tensile Strength for IN 792 DS from 20°C to 850°C, in the as received condition and after 5000h and 8540 h of thermal aging at 850°C.

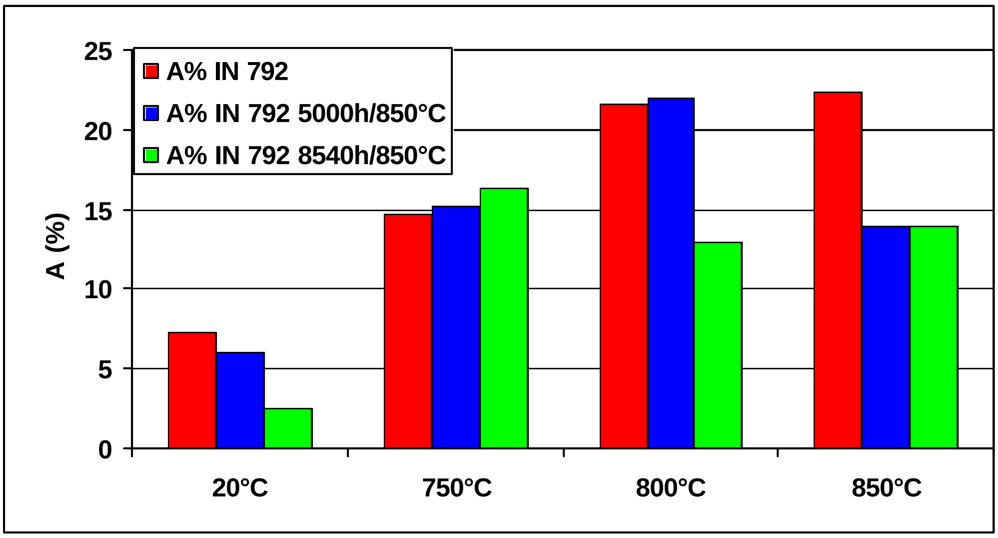


Figure 4-15 Elongation at rupture for IN 792 DS from 20°C to 850°C, in the as received condition and after 5000h and 8540 h of thermal aging at 850°C.

4.4.4 Thermal ageing heat treatment (850°C) on creep properties

Creep tests after ageing were performed at 850°C with the same stresses as for non-aged specimens.

Only creep tests on IN 792 DS were obtained after the 5000h/850°C heat treatment. The results from the two tests performed (850°C, 275 MPa) are presented below in Figures 4-16 and 4-17 together with a Larson Miller plot of IN 792 DS and CM 247 LC DS creep results.

A slight decrease in creep strength was observed, but the effect is not very strong. The detailed test results after ageing also show that the ductility is also not significantly reduced.

Figure 4-17 suggests that the effect of long term ageing on IN 792 DS (creep at 850°C/275 MPa) decreased the creep lifetime by a factor of 2, with a slight decrease of ductility. This suggests that IN 792 DS does not have a very stable structure at 850°C, and that the properties of the materials could be significantly altered by its long term use as turbine blades.

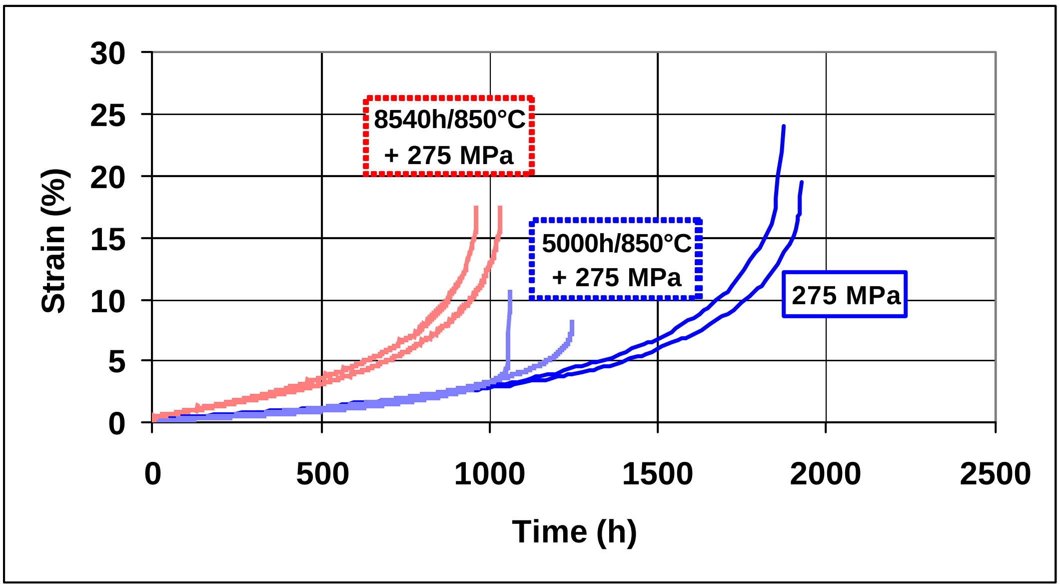


Figure 4-16 Creep curves obtained at 850°C/275 MPa for IN 792 DS specimens with and without prior thermal aging.

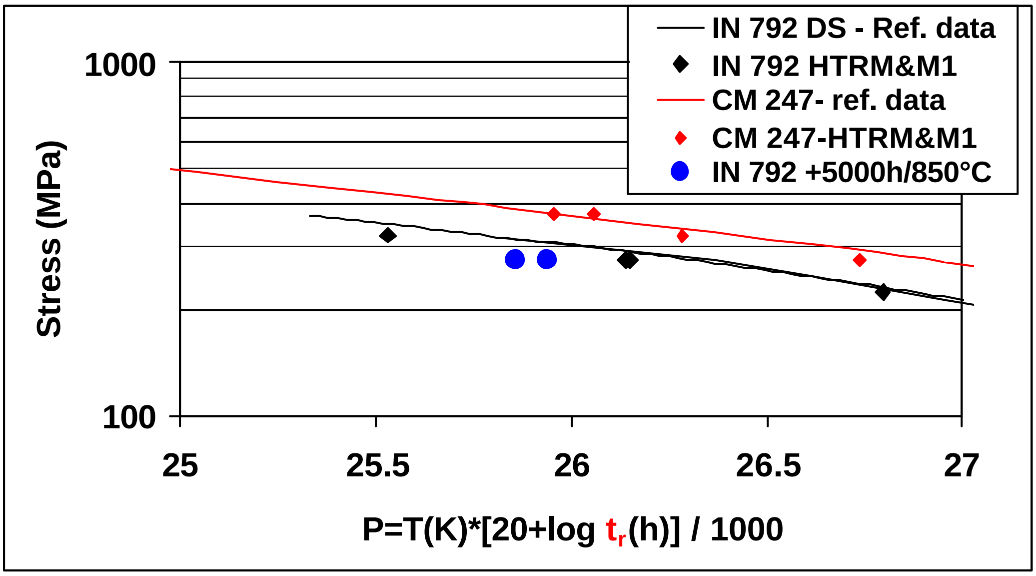


Figure 4-17 Results of CM 247 LC DS (as received) and IN 792 DS (as received and after 5000h/850°C HT) creep properties, with the Larson-Miller parameter calculated at rupture.

4.4.5 Fracture analysis

Examination of the fracture surface of a long term aged (850°C/8540 h) IN 792 DS specimen after creep testing at 850°C/275 MPa. Shows as with the un-aged specimens that the rupture mode remains ductile.

4.5 Summary Remarks

. Tests performed over a range of temperatures up to 850°C suggest a higher tensile and creep strength for CM 247 LC DS than IN 792 DS but with a lower ductility. Fitted curves using the Larson Miller parameter suggest that CM 247 LC DS may lose its strength at longer times and therefore longer term tests are needed at least 20000 h) to check this and also to compare with the Russian ISTC mechanical criteria for blades for the GTMHR.

The effect of thermally aging for up to one year at 850°C was investigated for the IN 792 DS material only. The results generally showed a slight decrease in tensile strength and high temperature ductility compared with non-aged material. The room temperature ductility however was severely reduced and because of this the material may be unacceptable for the turbine blade application. With regard to creep strength the results suggest a reduction (factor of approximately 2) and only a slight decrease in ductility. It will be necessary to perform some longer-term tests to see if the creep property decrease continues with longer times.

The effect of the applying a thermal cycle equivalent to the carburisation treatment performed at JRC (1000h/ 950°C) on the two blade material grades resulted in only a moderate effect on tensile strength. There is however an observed decrease in creep strength which will have to be taken into account when comparing with the properties of the carburised specimens (see below).

For these tests on untreated material and in air, the HTR-M1 results show that Directionally Solidified Nickel based alloys are potentially good candidates for the use as HTR turbine blades, however the long term stability of the alloy microstructure is a key point that needs further evaluated to ensure a safe use for the turbine.

5.0 RESULTS ON CARBURISED AND DECARBURISED SPECIMENS

For these tests HTR-M turbine and disc selected materials were pre-treated at two agreed extreme conditions – in the carburised and de-carburised conditions. The specimens from the as-received materials were "*pre-treated*" at a temperatures of 950°C in Ar + 2% methane for carburizing and in Ar + 500µbar H_2 + 50µbar H_2O for de-carburizing. These pre-treatments were specified by FZ Juelich based on their previous experience with similar materials [Ref. 2].

A special facility was built for these treatments at JRC. Unfortunately because of the unavailability of technician manpower the fabrication of this facility was seriously delayed. After completion some technical problems had to be solved before it was possible to pre-treat test specimens correctly. The facility consisted of a furnace with a 45mm diameter specimen chamber and a homogeneous temperature zone of 200mm. Appropriate gas mixtures could be fed through the specimen chamber. For the de-carburisation treatment a moisturiser was used to provide the required oxygen potential(Figure 5-1).

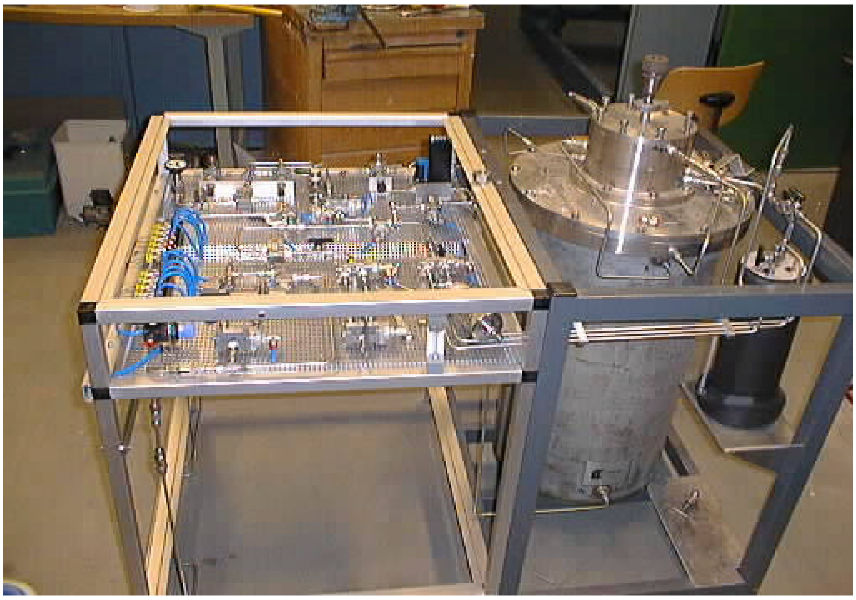


Figure 5-1 Carburisation Test Rig at JRC

In July 2004 the first dummy carburisation test on pieces of IN617 material from another project) was started. After the treatment the pieces were analysed by FZJ confirming that a carburisation depth of 0.9mm was achieved. Also some shrinkage porosity and internal oxidation of Al was observed in this layer.

The original planning had been to pre-treat specimen blanks and machine the specimens after the treatment, but because carburisation took place only on the surface machined specimens had to be treated instead. Even then the whole specimen cross-section would not be carburised. It was expected (and confirmed later) that a sufficient proportion of the cross-section would be carburised to allow the effect on creep properties to be observed.

5.1 Description of Pre-Treatments

The autoclave system for performing the carburisation (state III) and de-carburisation (state II) treatments was developed by the JRC and described in project Deliverable D4 [Ref 7]. All the specimens were treated in their as-machined form.

5.1.1 Carburizing (state III)

The conditions were as follows:

Gas composition:	Argon + 2%CH_4
Pressure:	0.5 bar
Flow rate:	120 ml/min
Temperature:	950°C

Duration: 1000 hours

The carburization treatment was completed for 5 specimens of IN 792 DS, 5 specimens of CM 247 DS and 5 specimens of Udimet 720.

IN792 DS: N1, N2, N3, N4, N5
CM247 LC DS: C1, C2, C3, C4, C5
U720: U9, U10, U11, U12

5.1.2 De-carburizing (state II)

The de-carburizing (state II) conditions were as follows
Gas composition: Argon + 500 mbar H_2
(to give 50 mbar H_2O with a dew point of -30°C)

- Pressure: 0.5 bar
- Flow rate: 120 ml/min
- Temperature: 950°C
- Duration: 1000 hours

The de-carburization treatment was completed for 5 specimens of IN 792 DS and 5 specimens of CM 247 DS:

IN792 DS: N6, N7, N8, N9, N10
CM247 LC DS: C6, C7, C8, C9, C10

It was found that the humidifier in the de-carburisation facility could not reach a low enough temperature to reach 50 ppm H_2O, and during the exposure the dew point increased from -30ºC to -11ºC. This meant that the humidity content exceeded the specified level, but this should enhance the de-carburization conditions.

5.2 Test Procedure/ facilities

The creep tests were performed with standard lever-arm creep machines using a three-zone resistance furnace system. The nominal test temperature of 850°C was monitored by Type S thermocouples attached close to specimen gauge length. The procedure and calibrations conform to ASTM E139-00.

To mount the button-headed specimens in the load train it became necessary to fabricate special adaptors since the loading bars on the machines were designed for specimens with threaded ends. The specimen elongation was measured by an extensometer system with two LVDTs. The reported values correspond to an average of the two signals.

The loads were selected to produce the following initial stress values for the two materials, which correspond to those used at CEA for the state 1 tests (non-damaged, as-received condition) and the Imod tests (thermal ageing treatment corresponding to the time-temperature conditions used during carburisation and de-carburisation):

IN 792 DS: 225, 275 and 325 MPa
CM 247 LC DS: 275, 325 and 375 MPa

5.3 Creep Test Results

Table 5-1 summarises the available results:

Table 5-1 Summary of creep test results available to date.

IN 792 carburised 950ºC, CH_4 120ml/min (dew point -30ºC) time1000h							
Specimen	Temp	Stress:	min. creep rate	$t_{1\%}$	t_f	A	Z
ID	°C	MPa	1/s	hrs	hrs	%	%
N1	850	505	5,50E-06	0,13	3,4	17,8	33,2
N2	850	325	8,30E-08	13,5	160,4	19,4	48,4
N3	850	275	2,50E-08	53,8	*Interrupted after 459 hrs.*	-	-
N4	850	225					

Material CM247 carburized 950ºC, CH_4 120ml/min (dew point -30ºC) time1000h							
Specimen	Temp	Stress:	min. creep rate	$t_{1\%}$	t_f	A	Z
ID	°C	MPa	1/s	hrs	hrs	%	%
C1	850	375	5,00E-08	26	295	28,2	38,9
C2	850	325	2,10E-08	41	792	32,8	40,5
C3	850	275	~7,40E-8	~190	*running*		

Material: IN792 decarburised 950ºC, Ar + 500µbar H2 = 50µbar H20 (dew point -30ºC) time 1000h							
Specimen	Temp	Stress:	min. creep rate	$t_{1\%}$	t_f	A	Z
ID	°C	MPa	1/s	hrs	hrs	%	%
N6	850	325	1,70E-08	86	840	19,1	43,9

5.3.1 IN-792 DS

Three tests were performed on the carburised specimens N1, N2, and N3. In the case of specimen N1, a mistake in measuring the specimen diameter lead to a much higher than intended stress and consequently a rupture time of only 3.4 hours. Also the test on specimen N3 was interrupted due to the failure of the adaptors between the specimen and the loading bars. New larger adaptors, using also a Nimonic 115 instead of Nimonic 105, are now being fabricated. The single test on the carburised specimen (N6) was competed successfully (Figure 5-2).

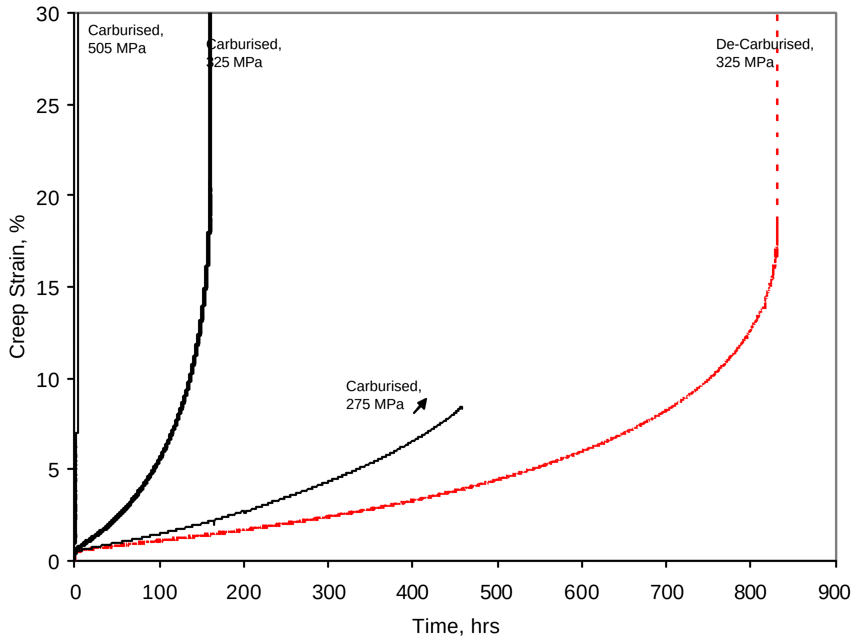


Figure 5-2 Creep curves for IN 792 DS specimens at 850°C.

The results for rupture time, minimum creep rate and elongation at failure are presented in Figures 5-3 to 5-5 and shown against the reference CEA tests in air. The results suggest that the carburised specimens produce lower creep rupture times and higher minimum creep rates in both the condition I (as–received) or I_{mod} conditions. However the creep ductility appears relatively unaffected.

In contrast the single test on a de-carburised specimen produced a slightly longer rupture time and slightly lower minimum creep than the I (as-received) condition.

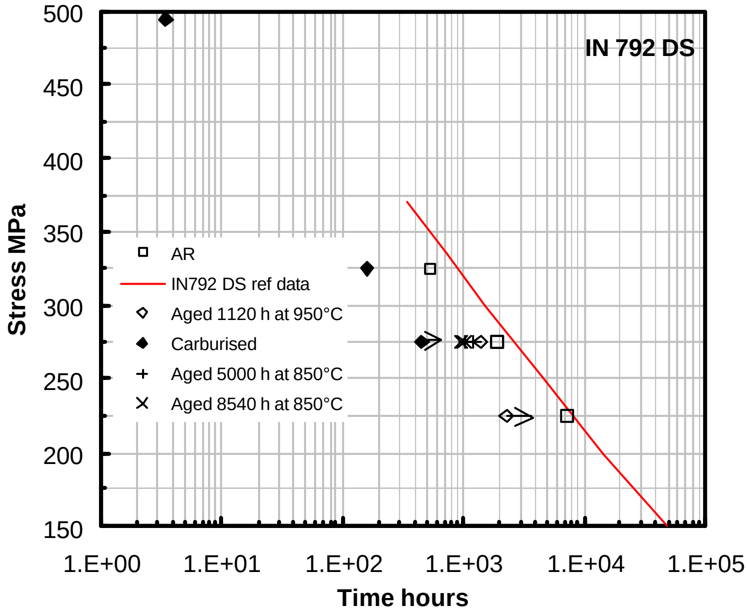


Figure 5-3 Creep rupture data for IN 792 DS at 850°C.

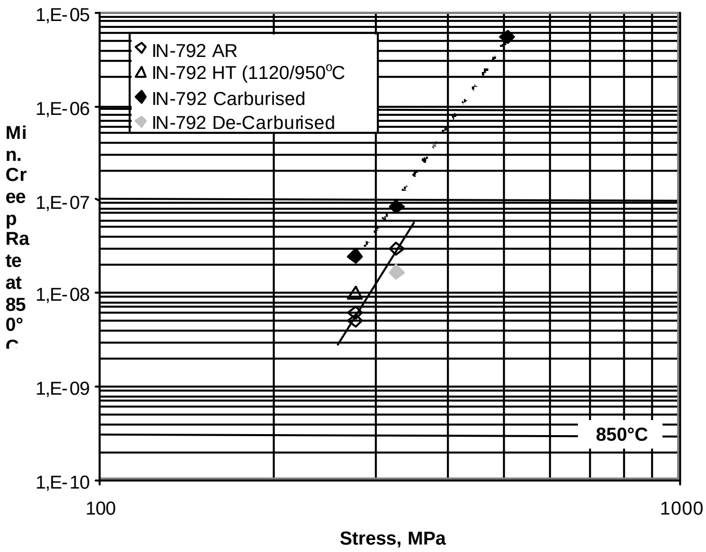


Figure 5-4 Minimum creep rate data for IN 792 DS at 850°C.

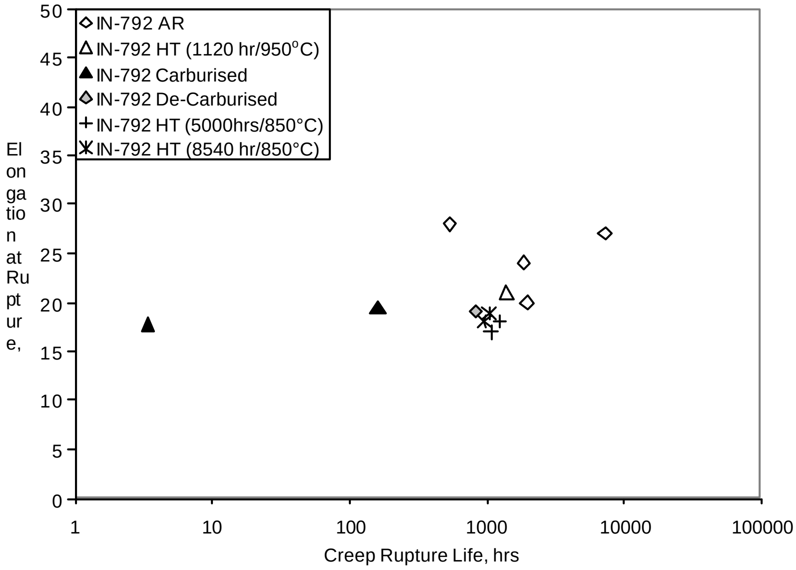


Figure 5-5 Elongation at rupture for IN 792 DS at 850°C.

5.3.2 CM 247 LC DS

Two tests were performed successfully on carburised specimens (C1 and C2) and a third is in progress. The creep curves from the competed tests are shown below in Figure 5-6. The results for rupture time, minimum creep rate and elongation at failure are presented in Figures 5-7 to 5-9 respectively. And against the CEA reference data. It is also evident in this case that the carburised specimens produce lower creep rupture times and higher minimum creep rates than those tested by CEA in either I (as–received) or I_{mod} conditions. However the creep ductility appears relatively unaffected.

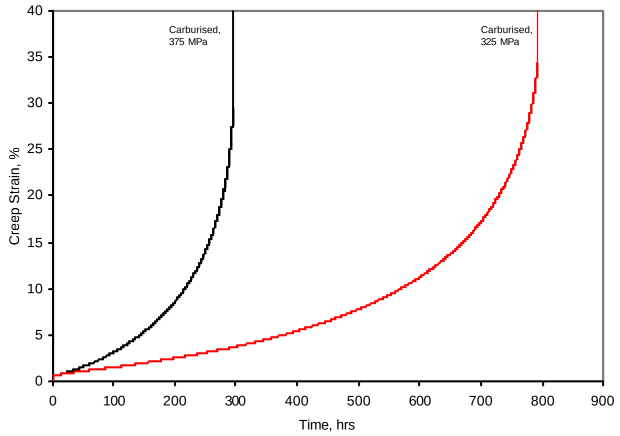


Figure 5-6 Creep curves for carburised CM 247 LC DS specimens at 850°C.

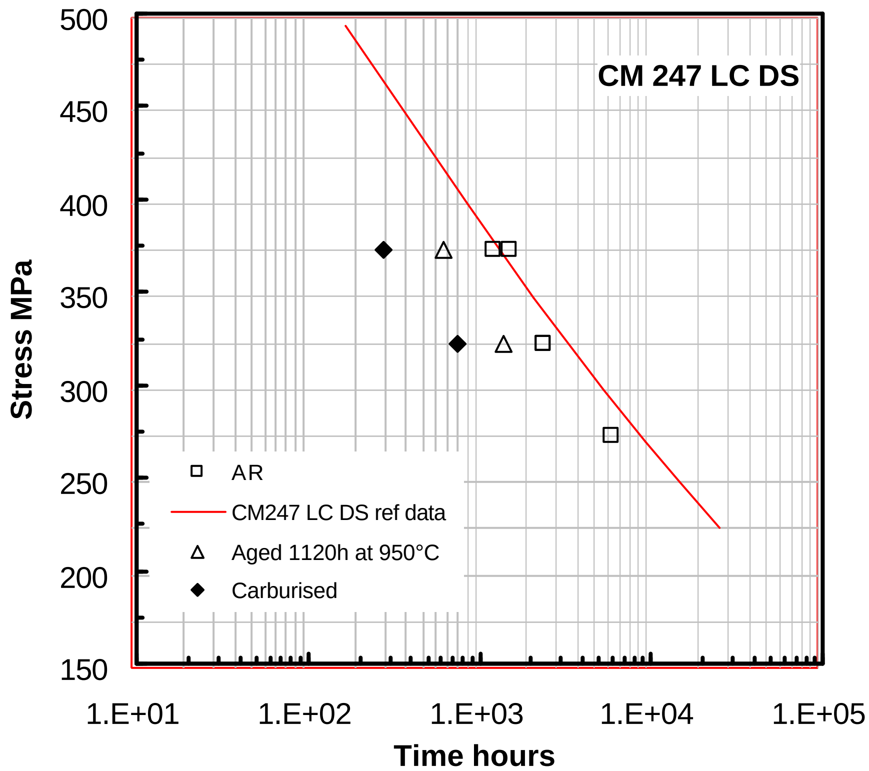


Figure 5-7 Creep rupture data for CM 247 LC DS at 850°C.

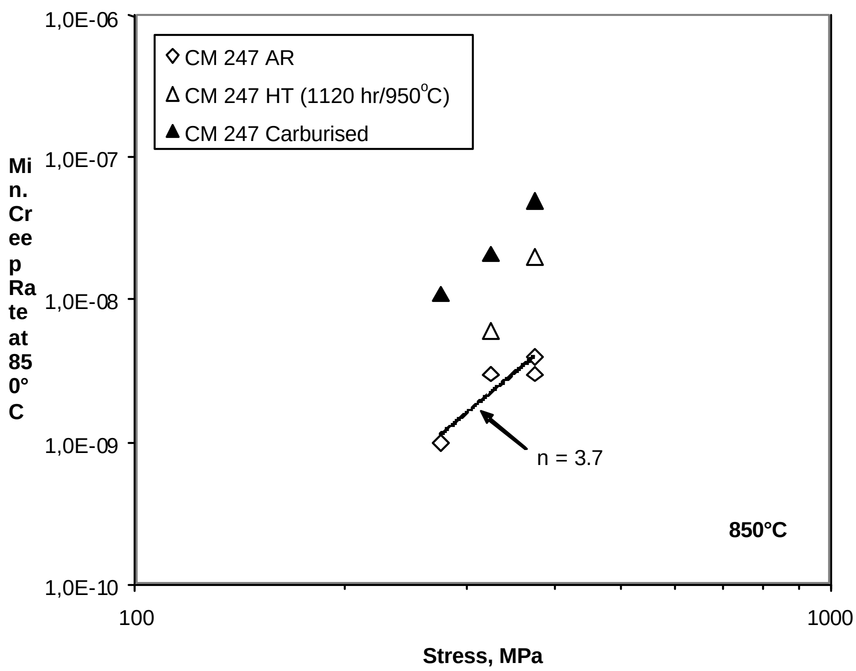


Figure 5-8 Minimum creep rate data for CM 247 LC DS at 850°C.

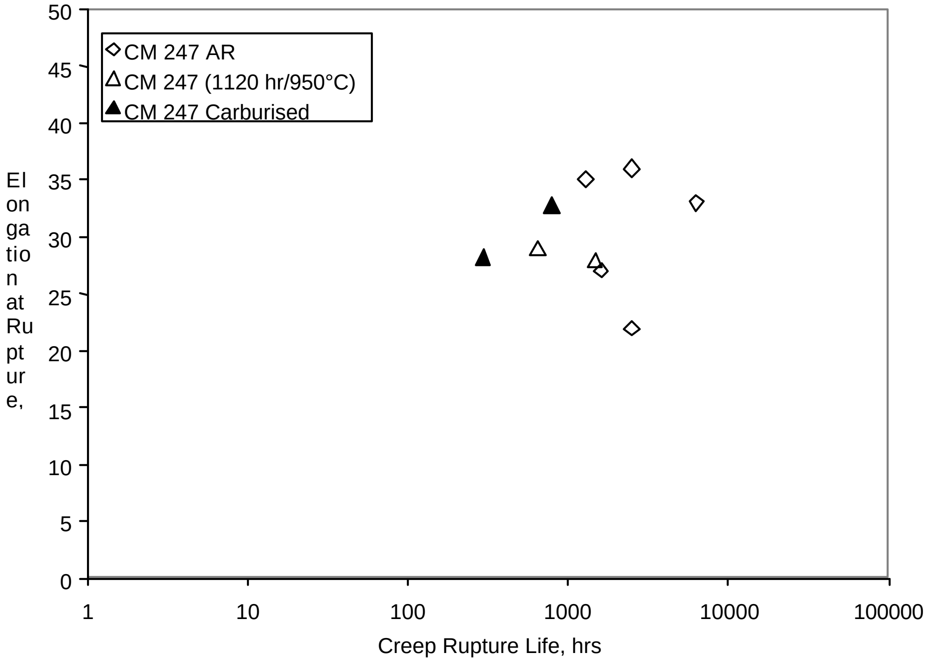


Figure 5-9 Elongation at rupture for CM 247 LC DS at 850°C.

5.3.3 Post-Test Microstructural Investigations

Post-test fractographic and metallographic examinations were performed on selected specimens to check the effect of the carburization and decarburisation treatments on the microstructure and to characterise the fracture mechanisms and surface damage effects. The investigations were done using a field emission scanning electron microscope (SUPRA 50) with secondary electron (SE) imaging mode. A step-wise procedure was followed using one half of the each specimen (the other half is kept intact): first the fracture surface was examined and then transverse and longitudinal cross-sections were cut, mounted and polished.

IN-792 DS

Fractographic and microstructural analysis was performed for specimens N2 (carburised) and N6 (de-carburised). Both were tested at 325 MPa/ 850°C, with rupture times of 160 and 840 hours respectively.

For the longitudinal cross-section of N2 (carburised) a band of dense carbides was seen at the sides of the specimen, extending to a depth of about 360 μm from the surface. This is also visible in the transverse cross-section. Further work is in progress to characterise the type of carbides present and the changes induced in the original microstructure. Creep damage was evident in the form of several cavities and pores. These appear mostly in the less carburised and un-carburised core regions, suggesting that the heavily carburised areas has good ductility at this temperature level. The damage is typically linked to the interfaces of the matrix with second phase particles. Since the creep tests were conducted in air, some surface oxidation occurred.

The longitudinal and transverse cross-sections of the de-carburised specimen N6 compared with the N2 carburised specimen suggests the creep damage is more evenly distributed across the section. Cavities and pores are typically linked to the interfaces with second phase particles. The surface oxidation damage is heavier that for the carburised specimen. An oxide layer and several oxidized surface cracks are also shown. A sub-surface chromium-depleted layer typically associated with such high temperature oxidation was identified.

It should be pointed out that although the creep rupture ductility value was similar for both specimens, the carburised N2 showed a more mixed range of ductile and brittle fracture features than the de-carburised specimen. In the case of de-carburised N6, ductile features were pre-dominant.

CM 247 LC DS

The analysis on specimens in CM 247 LC DS has been limited so far to fractography of the carburised specimen C2, which was also tested at 325 MPa/ 850°C and produced a rupture time of 792 hours. The fracture features are predominantly ductile, as expected.

5.4 Low Cycle Fatigue tests on Turbine Disc Material Udimet 720

These tests were planned to be completed within the HTR-M Programme. However due to delays in the commissioning of the corrosion test rig as described above they are reported here as part of the HTR-M1 results.

The material Udimet 720 stems from a forged pancake (certificate no. HY540203), which CEA had received from Aubert & Duval. One half of the pancake disk (diameter 320mm, height 80mm) was sent to JRC-IE in the as received state (cast & wrought, forged and fully heat treated for service), henceforth referred to as State I, for specimen machining and further carburization treatment. This treatment was carried out at JRC-IE on six fully machined LCF specimens by exposing them to a 1.6 bar absolute Argon+2%CH_4 atmosphere for 1000h at 950ºC. This gave rise to a heavily carburized material state, henceforth called State III. Two specimens were left in the as-received State I for reference tests and for temperature profile calibration.

The carburization treatment had resulted in pitting corrosion visible at low magnification on the specimen surface. It was therefore decided to re-polish three specimens according to ASTM E606-92, while leaving the other three specimens in the as exposed state.

5.4.1 Experimental Details

All Low Cycle Fatigue (LCF) tests were carried out at a temperature of 650°C. The temperature was achieved within a gauge length of 8mm using induction heating, maintained within +/- 2°C by controlling the temperature by a Type N thermocouple, spot-welded 6mm below the specimen centre (i.e. 2mm outside the gauge length), but still within the parallel length. For the purpose of these tests, the extensometer was re-calibrated to Class1, and the load train misalignment verified to be less than 3% bending percentage by the use of a strain-gauged alignment specimen.

The actual specimen diameters and actual gauge lengths of the Sandner clip-on extensometers were determined individually to 10µm accuracy (3 repeat measurements). Room temperature Young's modulus was measured on two opposite specimen sides to check for appropriate mounting of the specimen and the extensometry. Young's modulus and yield stress ($Rp_{0.2\%}$) at 650°C were determined from the first cycle of the LCF test.

All LCF tests were performed in total strain control, using triangular strain waveforms, a strain ratio of $R\varepsilon$ = 0.05 (always tension), and different maximum strain values of 0.6%, 0.7%, 0.8%, and 1.0%. The cycle period was varied in order to maintain a constant strain rate of +/- 0.1%/s for all tests. The tests were conducted up to a 15% drop of maximum load or failure or to N ~ 45,000 (termination), whichever occurred first. Details of the test Matrix are given in Table 5-11 below:

Table 5-2 Test Matrix for U 720 Carburised test specimens

Specimen id.	state	T (°C)	de/dt (10^{-3}/s)	t_{cycle} (s)	e_{max} (%)	e_{min} (%)
U1	III	650	1	19.0	1.0	0.05
U2	III	650	1	13.3	0.7	0.035
U3 (polished)	III	650	1	15.2	0.8	0.04
U4 (dummy)	I	Temperature profile calibration				
U5	III	650	1	11.4	0.6	0.03
U6 (polished)	III	650	1	19.0	1.0	0.05
U7 (polished)	III	650	1	13.3	0.7	0.035
U8	I	650	1	19.0	1.0	0.05

5.4.2 Results

Table 5-12 summarizes the main results form the LCF Fatigue tests. One specimen was lost due to a power trip leading to specimen failure by thermal contraction. The other tests were considered valid with fatigue cracking occurring within the parallel length. One specimen (U8) failed close to, but not immediately at, the thermocouple spot-weld, where the temperature was about 4°C lower. All other specimens failed within the gauge length, where the temperature deviation was less than 3°C.

It is apparent from Table 5-12 and Figures 5-1 and 5-2 that the carburization treatment has detrimental effects on the yield strength (σ_y = 910 MPa for State I, whereas $\sigma_y \cong$ 650 MPa for State III), also on the LCF performance, particularly in the regime of high strain ranges and low numbers of cycles to failure. For a given strain range of 0.95% (i.e. max. strain 1%), the State I material bears significantly larger stress ranges while LCF lives are still longer. This observation holds irrespective of the surface quality (corrosion pitted, Fig 5-2 or repolished Fig. 5-1).

Table 5-3 Overview of U720 LCF test results

	LCF condition		**1st cycle**		**Test end**		***Nf/2***				
Spec. id	ε_{max} **(%)**	ε_{min} **(%)**	**E (650°C) (Gpa)**	**Rp0.2% (650°C) (Mpa)**	N_f	**Terminated by**	σ_{max} **(Mpa)**	σ_{min} **(Mpa)**	$\Delta\varepsilon_{tot}$ **(%)**	$\Delta\varepsilon_{pl}$ **(%)**	$\Delta\varepsilon_{el}$ **(%)**
U1	1.0	0.05	174.5	648	647	brittle fracture inside GL	726.4	-675.7	0.950	0.157	0.804
U2	0.7	0.035	169.3	652	3353	brittle fracture inside GL	696.3	-399.1	0.665	0.024	0.647
U3	0.8	0.04	Specimen went out of control due to general power failure in North Holland								
U5	0.6	0.03	169.5	640	44694	Stopped by operator	638.9	-289.8	0.570	0.024	0.549
U6	1.0	0.05	170.1	650	618	15% load drop crack inside GL	731.6	-651.9	0.953	0.147	0.813
U7	0.7	0.035	168.1	628	45549	Stopped by operator	673.1	-410.9	0.664	0.029	0.64
U8	1.0	0.05	169.1	910	3392	15% load drop crack outside GL	936.3	568.3	0.953	0.068	0.890

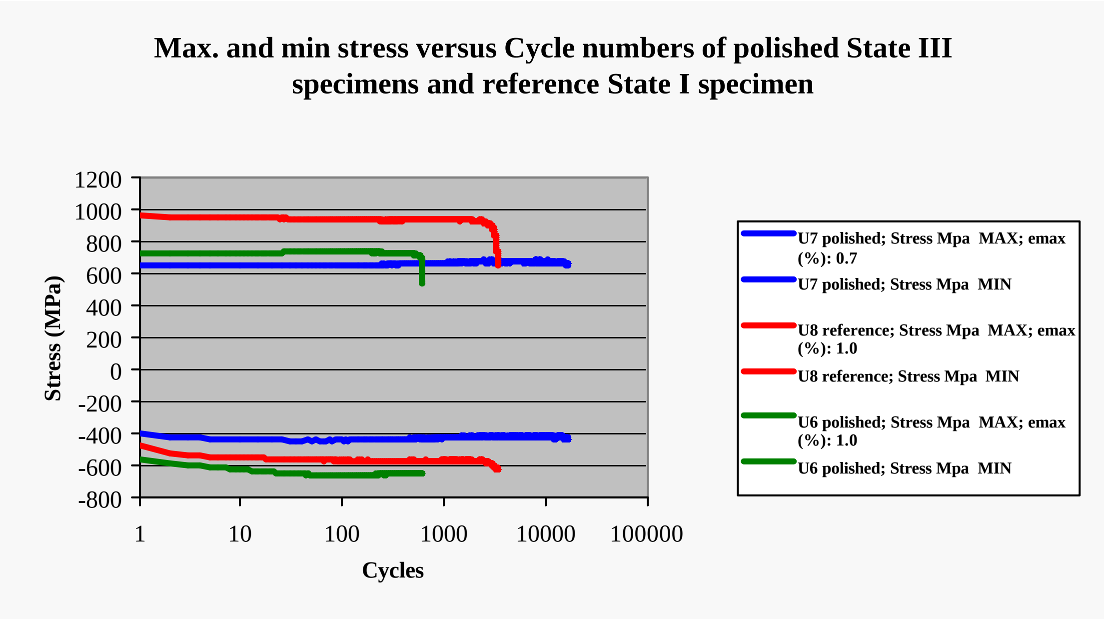


Figure 5-10 Cyclic stress maxima and minima of polished State III (e_{max} = 0.7% and 1.0%) as compared to reference State I (e_{max} = 1.0%)

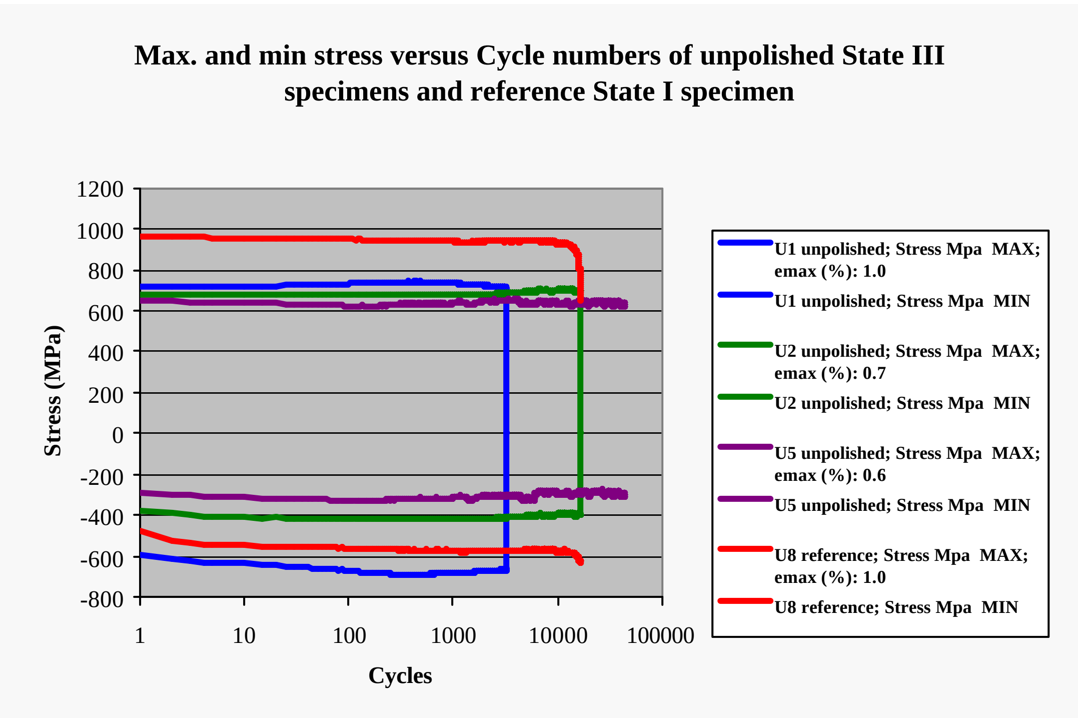


Figure 5-11 Cyclic stress maxima and minima of unpolished State III (e_{max} = 0.7% and 1.0%) as compared to reference State I (e_{max} = 1.0%)

A more comprehensive check is obtained when taking the results from the CEA tests in air (Reference. 8) – Deliverable D2 into account – Fig 5-3. It should be noted that a check LCF test was performed at JRC (using ε_{max} = 1%) to make sure the similar results were obtained using the JRC facility and the result fits convincingly into the CEA series of results.

The comparison suggests that the numbers of cycles to failure are reduced by a factor of about 5 for State III as compared to State I, if maximum strains ε_{max} > 0.7% are considered. For these conditions re-polishing of the carburized test pieces has little effect, as fatigue lives tend to be bulk controlled. It is expected that surface finish will have more of an effect at low strain range levels, since fatigue crack initiation (which predominates in this regime) starts from surface defects, such as corrosion pits.

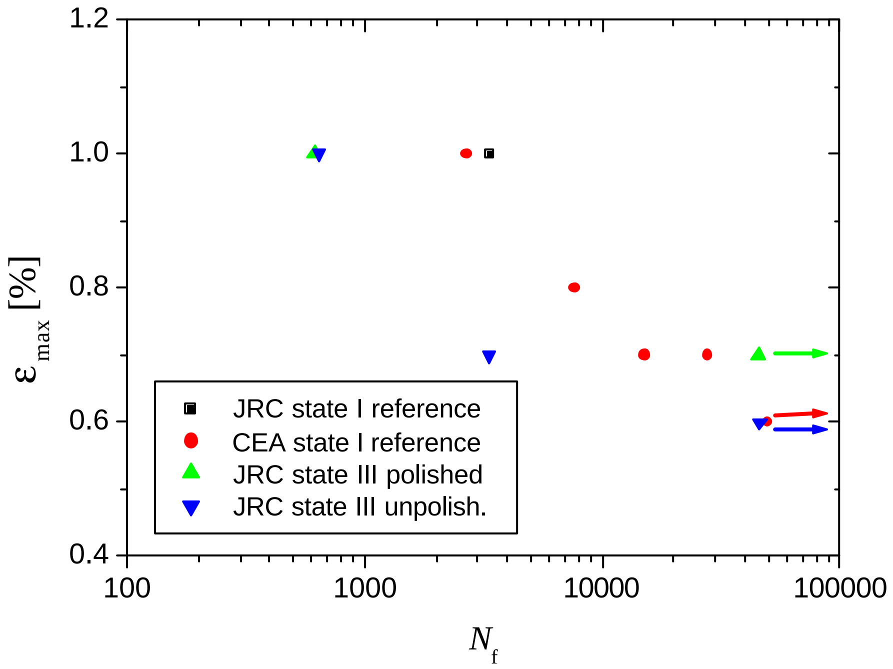


Figure 5-12 Maximum total strain versus number of cycles to failure, incl. CEA results for reference State I

Taking the plastic strain ranges at midlife ($N_f/2$) against N_f, (see Figure 5-4) suggests that in comparison with the State I (as received) condition, the State III (carburised) material can accommodate much larger plastic strains and still survive for more than 45000 cycles. This holds for the lowest max. strain values and, in particular, the polished State III.

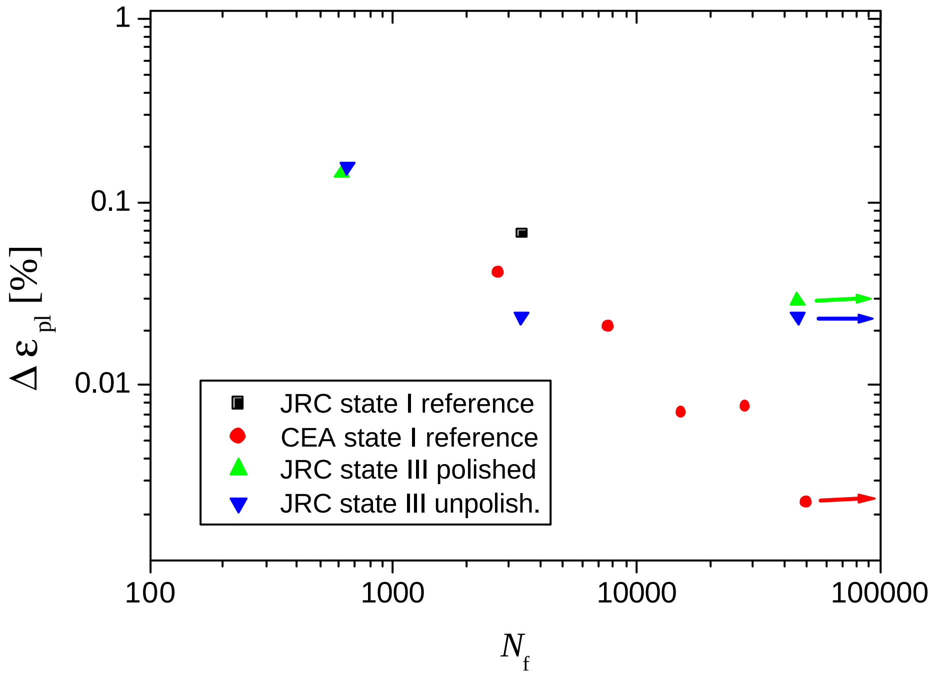


Figure 5-13 Plastic strain range versus number of cycles to failure, incl. CEA results for reference State I

5.5 Summary Remarks

The results of the short term creep tests performed so far show that the carburisation treatment causes a marked loss of creep strength and increase in minimum creep rate with respect to the as-received condition (state I) and to the thermal aged condition (state I_{mod}) for both alloys. The rupture ductility remains comparable to that of the as-received condition. The reduction factor on creep rupture appears to be of the order of 5-10 on life however further tests are needed to confirm this.

On the evidence of the few results available CM 247 LC DS appears to give better creep rupture properties than IN 792 DS in the carburised condition.

The single result for IN 792 DS in the de-carburised condition suggests there is no noticeable reduction in creep life.

Carburization results in approximately a 30% drop of yield strength.

Carburization results in up to five times shorter low cycle fatigue (LCF) lives albeit stress ranges are lower.

Further micro-structural analysis is needed to:

- assess the fracture mode (optical microscopy),
- determine the penetration depth of carburization treatment (SEM),
- clarify the role of persistent slip bands in fatigue crack initiation and failure (TEM).

Further testing is continuing:

- On the remaining available specimens in the HTR-M1 Programme to help clarify the reduction factors for carburisation and confirm the creep behaviour of the two selected blade materials. These will be reported in the Raphael-IP within the 6th Framework Programme.

Further testing is needed:

- To check the de-carburized condition with respect to creep and LCF and to arrive at a more comprehensive understanding of the effect of this condition on behaviour.

6.0 CONCLUSIONS & RECOMMENDATIONS

6.1 As received condition

Within ther scope of the tests performed, CM 247 LC DS shows the best mechanical properties (tensile, creep) at high temperature. It is further recommended however to carry out longer term creep tests (at least 20000 h) to confirm its suitability. Such tests would allow a more comprehensive check against the ISTC mechanical criteria for blade materials (60,000h design life) to be made.

6.2 Heat treated and Aged

A thermal aging of nearly one year at 850°C leads to a slight decrease of tensile strength and high temperature ductility for IN 792 DS, however room temperature ductility is severely reduced. A decrease in creep strength was observed after aging and further tests are needed to confirm whether a longer aging time leads to even further reductions.

The long term stability of the Directionally Solidified nickel based alloy microstructure needs to be further evaluated to ensure a safe use for the turbine.

The heat treatment caused by the thermal cycle used for the carburisation treatment performed at JRC (1000h/950°C) has a moderate effect on tensile strength for the two grades. A decrease in creep strength was also apparent which has to be considered when comparing with the results from the carburised specimens.

6.3 Carburised specimens

The carburization treatment on the IN-792 DS and CM247 LC DS alloys produced a densely carburised zone to a depth of approximately 360 µm from the surface and a substantial increase in carbides also in the core of the specimens.

Carburization results in approximately a 30% drop of yield strength and lowers the creep rupture strength of the materials. The reduction factor is of the order of 5-10 reduction in life, but this has to be confirmed by further tests.

On the evidence of the few results available CM 247 LC DS appears to give better creep rupture properties than IN 792 DS in the carburised condition.

The effect of carburization on LCF tests on in 792 DS results in up to five times shorter lives albeit with stress ranges that are lower.

Further testing is continuing on the remaining available specimens. Additional tests and further micro-structural evaluations are needed to re-enforce the findings from these few available tests.

6.4 De-Carburised specimens

The single test to date on a de-carburised IN-792 DS specimen produced a rupture life comparable to that obtained for the as-received condition. Additional creep and fatigue testing is needed to understand the effect of this condition on behaviour and to arrive at a more comprehensive understanding for design interpretation.

6.5 Remaining testing

Execution of the remaining creep tests foreseen in the original HTR-M1 programme is in progress, as are micro-structural investigations to characterise the effects of carburisation and de-carburisation on these alloys. The results will be reported to the RAPHAEL project.

7.0 REFERENCES

1. D. E. Buckthorpe
HTR-M Final Report, NNC Report No. 68571/TR/0008
HTR-M 04/12 P 0 0 105, December 2004

2. H. Rantala

Notes of a HTR-M & M1 Specialist Meeting on Corrosion & Carburisation, HTR-M 03/07 M 2 0 72

3. D.E.Buckthorpe
Minutes of the HTR-M/ M1 Progress Meeting on 15-17 April 2002, Julich, Germany, HTR-M1 02/5 M 0.0.5

4. R. Couturier, J Calapez, N. Scheer,H. Rantala
Materials for Turbine: Final Report on results from HTR-M investigations, HTR-M 04/10 D 2 3 103

5. ASM 1997

6. A. Romantsov
Final Report for ISTC Project 1313-2001. Testing and Investigations of Materials for Turbo-compressor High Temperature Components
HTR-M1 05/01 D 1 0 45

7. R Couturier, J Calapez, H Rantala
HTR-M1 Deliverable D4: Final HTR-M1 Report on HTR blade alloy properties. HTR-M1 05/10 d 1 2 35

8. R Couturier, H. Rantala
HTR-Ml Deliverable D2: HTR-M1 Report:-Intermediate report on Material Testing of Turbine Materials, Doc. HTRM1-05/07 D l 2 30.

T A B L E S

Table 0-1: HTR-M1 Deliverables

Deliverable No	Deliverable title	Delivery date	Current planned	Date Achieved	Main Organisation
D1	Report on selection of materials & testing requirements	12	01/10/2002	11	CEA
D2	Report on specimen fabrication and procurement & testing requirements	18	01/04/2003	17	CEA
D3	Intermediate report on material testing	47	31/07/2005	47-	CEA
D4	Final report on blade alloy properties	51	31/12/2005	51	CEA
D5	Review of results and updating of data base	51	31/012/2005	Suppressed not needed	CEA
D6	Final Report summary	51	31/12/2005	-51	NNC

Table 0-2 HTR-M1 Issued Scientific Reports and Minutes on Turbine Materials Development

HTR No.	Title	Author / Org
HTR-M1 02/1 M.0.0.2	Minutes of the HTR-M1 Kick-off plus 2nd HTR-M Progress Meeting on 5-7 December 2001, Grenoble	D.E. Buckthorpe/ NNC
HTR-M1 02/5 M 0.0.5	Minutes of the HTR-M / M1 Progress Meeting on 15-17 April 2002, Julich, Germany	D.E. Buckthorpe/ NNC
HTR-M1 02/9 D 1 1 7	HTR-M1: Report on selection of materials & testing requirements for the Turbine	**R. Couturier / CEA**
HTR-M1 02/11 P 0 0 8	Mid-term report for the HTR-M & M1 Projects	D. E. Buckthorpe/# **NNC**
HTR-M1 02/12 M 0 0 10	Minutes of the HTR-M plus M1 Mid Term Meeting on 28-29 November; 2002, Framatome, Paris	D. E. Buckthorpe NNC Ltd
HTR-M1 03/03 D 1 0 12	Report on specimen fabrication, procurement & testing requirements for the turbine blade material	R. Couturier / CEA
HTR-M1 03/07 M 0 0 17	Minutes of the 5th HTR-M plus M1 Progress Meeting	D. Buckthorpe, R. Bestwick, A. A. Braithwaite / NNC
HTR-M1 03/11 P 0 0 18	36 Month HTR-M plus 24 month HTR-M1 Management Report	D. E. Buckthorpe/ NNC

TABLE 2 (CONT):

HTR No.	**Title**	**Author / Org**
HTR-M1 03/11 P 0 0 19	Third Annual Technical Report for the HTR-M project	D.E. Buckthorpe / NNC, et al.
HTR-M1 03/12 M 0 0 20	Minutes of the 6th HTR-M plus M1 Progress Meeting at FZJ	D. E. Buckthorpe/ NNC
HTR-M1 04/07 M 0 0 23	Minutes of the 7th HTR-M plus M1 Progress Meeting at EA, Madrid	D. Buckthorpe/ NNC
HTR-M1 04/11 M 0 0 24	Minutes of the 8th HTR-M plus M1 Progress Meeting at Framatome, Lyon	D.Buckthorpe/ NNC
HTR-M1 04/10 P 0 0 26	3rd Scientific Report on HTR-M1 Project	D. Buckthorpe/ NNC
HTR-M1 05/05 M 0 0 29	Minutes of the 7th HTR-M1 Progress Meeting held at NNC, Knutsford	D.Buckthorpe/ NNC
HTR-M1 05/07 D 1 2 30	HTR-M1 Report:-Intermediate report on Material Testing of Turbine Materials	R. Couturier/CEA H. Rantala/JRC
HTR-M1 04/11 D 0 0 31	Technical Implementation Plan for the HTR-M1 Project	D.Buckthorpe/ NNC
HTR-M1 05/03 P 0 0 32	42nd Month HTR-M1 Management Report	D.Buckthorpe/ NNC
HTR-M1 05/10 M 0 0 33	Minutes of the Final HTR-M1 Progress Meeting held at NRG, Petten	D.Buckthorpe/ NNC
HTR-M1 05/11 D 1 2 34	Report on results of heavily carburised U720 disk material	S Ripplinger, H Rantala, P Hahner/ JRC-IE
HTR-M1 05/11 D 1 2 35	Final HTR-M1 Report on HTR blade alloy properties	R Couturier, J Calapez/ CEA H Rantala/ JRC-IE
HTR-M1 05/10 D 0 0 36	Presentation of results for VHTR – results from HTR-M & HTR-M1 Projects plus future programme, EUROMAT Conference, Prague, Czech Republic, 5–7th September 2005	D. Buckthorpe/NNC et al
HTR-M1 05/03 P 0 0 37	Request for HTR-M1 Contract Extension from the Partnership	D. Buckthorpe et el
HTR-M1 05/03 P 0 0 38	Approval of HTR-M1 Contract Extension from the Commission	P. Fernandez Ruiz/ EU
HTR-M1 05/12 D 1 2 43	HTR-M1 Turbine Results data provided on CD-Rom	D Buckthorpe/NNC
HTR-M1 05/12 D 1 2 44	HTR-M1 Turbine Materials – Final Summary Report	D. Buckthorpe/NNC
HTR-M1 05/01 D 1 0 45	Final Report for ISTC Project 1313-2001. Testing and Investigations of Materials for Turbo-compressor High Temperature Components	A. Romantsov
HTR-M1 05/12 D 0 0 46	Final Report for HTR-M1 Project	D. Buckthortpe/NNC

Table 0-3 HTRM1 Turbine Material Test Matrix

CM 247 at 850°C:

Test/no of specimens	T (°C)	I	I_{mod}	II	III
TENSILE	20 750 800 850	1 1 1 1	1 1 1 1		
CREEP X Mpa (3000 h) Y Mpa (1000 h) Z Mpa (5-10 Kh)	850	2 2 2	2 2 2	2 2 2	2 2 2

x., y, z : stress level to be defined to attain the desired rupture time.

For IN 792 DS aged up to 8000 H at 850°C:

Test	T (°C)	I	I_{mod}	II	III
TENSILE	20 800 850	1 1 1			
CREEP **x MPa (1000h)** **y MPa (3000h)**	850	3		4	4

x and y: stress level to be defined to attain the desired rupture time.