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Archer
Advanced Reactor
for Cogeneration of
Heat & Electricity-R&D

Deliverable D-43.51: Review of existing data on Alloy 800H (focusing on thin sheet) and produce results applicable to mock-up and IHX.

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Summary

This document contains the results of a preliminary review of available data on Alloy 800H. The results have been made available to the project by AREVA-NP and ALSTOM.

In the context of the IHX mock-up being prepared within WP4.3, no evidence has so far been found that the design properties of Alloy 800H are significantly reduced for thin sections and in impure helium. However, there is sufficient information from other studies to suggest that further work may be necessary to validate the properties of thin-section Alloy 800H prior to the design of the IHX for the demonstration and subsequent plant.

It is therefore recommended that the available material design data provided by AREVA is used for the analysis and assessment of the mock-up. The material design data may also be used for the preliminary analysis and assessment of the IHX, but may require revision in the light of further studies on thin-section material, and the effect of impure helium.

The material properties for thin-section material will be kept under review within the project, and may be changed should results from later investigations show otherwise.

Approval

Rev.	Date	First authors	SP leader	Project Coordinator
0	month/year	S DUBIEZ-LE-GOFF C BULLOUGH	D Buckthorpe; AMEC	S Groot;NRG
	8/2011			

Distribution list

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All beneficiaries	ARCHER	Through internet workspace
WP4 Participants	ARCHER	Through internet workspace

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1 Introduction

This document contains the results of a review of available data on Alloy 800H. The results have been made available to the project by AREVA-NP and ALSTOM.

Although a significant amount of material test and property data are available for thicker section Alloy 800H (typically plate or bar material exceeding 15mm section), the amount of testing on Alloy 800H or other variants of Alloy 800 in sections of 5mm or less is low. Within this report results are presented on Alloy 800H design properties and thin and thick section properties that has been found,

2 Summary of Contents

Contribution from AREVA-NP

Results are presented for Alloy 800H design data. The properties included are those for:

- Youngs' Modulus
- Thermal Expansion Coefficient
- Yield Strength
- Stress to Rupture and minimum creep rate
- Fatigue curves

Results are presented in the form of temperature varying properties as tables and charts. Where applicable comparisons have been made between EN NF, Draft KTA and ASME Code values and recommendations made regarding which data source should be used..

The corrosion of Alloy 800H under HTR helium environment means that its application is limited to temperatures below 850^{o}C

Welding coefficients provided within the KTA Draft Standard are not considered representative of laser welding which is to be used for the IHX Mock-up fabrication.

Contribution from ALSTOM

The focus of this review has been to search for available thin section properties and behaviour.

Sheet material in the range of 3-5 mm section can be produced by two related production methods: hot-rolling and cold-rolling. The significance of these processes on material properties and behaviour has been summarised.

With regard to component manufacture cold bending, stamping and/or other cold forming operations may affect properties, also welding or brazing. Excessive machining rates can also lead to work hardening at surfaces and the buildup of residual stresses and possibly some local re-crystallization.

Exposure to impure helium can lead to carburisation or decarburisation affecting properties and strength.

A preliminary review of the key material properties was made including tensile and creep rupture strength. Properties associated with fatigue are still to be investigated and summarised.

3 Conclusions

The main conclusions arising from both investigations are as follows:

Design properties are available for Alloy 800H from existing Codes and Standards sources and where appropriate these are to be used unless there are other specific information on thin section properties.

The temperature limit for the material for IHX application is considered to be 850^{o}C.

Weld factors for laser welding are not yet available but may come from further work or specific tests on welded thin sections to be carried out later in the project.

A preliminary review suggests design properties for the thinnest sections may not be affected by the manufacturing process. This is dependent on the machining process.

Cold working operations should be kept low and at least below 20% to minimise re-crystallization effects. This limit will be further reviewed when more information becomes available.

Issues such as weldment relaxation, metal dusting are unlikely to have a significant influence on properties and behaviour. Laser welding will be used for the manufacture and carbon activity levels will be low.

Effects of helium environment on properties will require further investigation and little tensile or creep data for thin sections have so far been found that is relevant to the IHX conditions.

The effect of grain size and section thickness on creep properties may be important and in extreme cases strength properties may be reduced by up to 15%.

4 Recommendations

In the context of the IHX mock-up being prepared within WP4.3, no evidence has so far been found that the design properties of Alloy 800H are significantly reduced for thin sections and in impure helium. However, there is sufficient information from other studies to suggest that further work may be necessary to validate the properties of thin-section Alloy 800H prior to the design of the IHX for the demonstration and subsequent plant.

It is therefore recommended that the available material design data provided by AREVA is used for the analysis and assessment of the mock-up and IHX.

This situation will be kept under review within the project and may be changed should results from later reviews show otherwise or the specific planned tests on thin section sheet and weldments show the strength properties to be significantly affected.

CONTRIBUTION FROM AREVA

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Titre du document

ARCHER: 800H Design data

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RÉVISIONS

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SUMMARY

In the frame of ARCHER European project, this document aims at supplying the Alloy 800H data needed to perform design analyses for the Heat Exchanger up to 800°C:

- Young modulus
- Thermal expansion coefficient
- Yield strength
- Stress to rupture and minimum creep rate
- Fatigue curves

Some elements on the behavior under High Temperature Reactor environment will be given.

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0. REFERENCES

[1] Norme NF EN 10302
Creep resisting steels, nickel and cobalt alloys– May 2008

[2] ASME SB 163 - ASTM B 163-04
Specification for seamless nickel and nickel alloy condenser and heat-exchanger tubes.

[3] ASME SB 407 - ASTM B 407-04
Specification for nickel-iron-chromium alloy seamless pipe and tube

[4] ASME SB 408 - ASTM B 408-04
Specification for nickel-iron-chromium alloy rod and bar

[5] ASME SB 409 - ASTM B 409-04
Specification for nickel-iron-chromium alloy plate, sheet and strip.

[6] ASME SB 564 - ASTM B 564-06
Specification for nickel alloy forgings

[7] 2010 ASME Code, Section III Subsection NH

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1. INTRODUCTION

1.1. 800H grade presentation

The alloy 800H is an alloy of about 32% nickel and 21% of chromium. Its iron content reaches almost 50%, it could be considered as an austenitic stainless steel with 32% of nickel. For wrought products, the chemical composition is defined in Europe by the standard EN 10302 (Ref.1), its designation X8NiCrAlTi32-21 (1.4959) being that of a steel. This composition is given in Table 1 as well as those specified by ASTM standards for the grade UNS N08810 (official designation of the alloy 800H) and by the German draft code KTA (Reference 13) for the grade 800H. This composition indicates intentional additions of titanium and aluminum, as minimum levels required for those two elements.

The metallurgical condition of the standard EN 10302 is designated by + AT. This is a solutionning in the temperature range 1100-1200°C followed by air cooling or accelerated. Although water is not specified as a coolant, this treatment is like an annealing.

The 800H designation (UNS N08810) is ambiguous because, by analogy with the sub grades H of austenitic stainless steels, it denotes a carbon content more than 0.05%. However, it also differs from the Alloy 800 designation (UNS N08800) since it has an austenitising treatment leading to a larger grain size and improved creep properties. A third designation, Alloy 800HT (UNS N08811) is essentially a subset of Alloy 800H with further improvements in creep properties.

1.2. Codes and standards covering 800H alloy

1.2.1. European Standard

The following mechanical properties are specified at room temperature in the European Standard (Ref. 1):

- Yield strength, $R_{p\ 0,2} \geq 170$ MPa
- Tensile strength, 500 MPa ≤ R_m ≤ 750 MPa,
- Strain to rupture A ≥ 35 % for flat products; A ≥ 30% for long products.

There is no European Standard for forgings.

Alloy 800H is not considered by any applicable European code for pressure application. But the document KTA in Reference 13 provides data at high temperatures and very high temperature for three sub-grades of Alloy 800: for two of them, the specified minimum values for Rp 0.2, Rm and A% are consistent with those of European standard. For the third, that does not have the metallurgical state AT, a minimum value of 210 MPa is required to Rp 0.2.

1.2.2. American Standard ASME/ASTM

The following ASME/ASTM standards 800H alloy:

- ASME SB 163 - ASTM B 163 (Ref.2), seamless tubes for heat exchanger,
- ASME SB 407 - ASTM B 407 (Ref.3), seamless tubes,
- ASME SB 408 - ASTM B 408 (Ref.4), bars and rods,

- ASME SB 409 - ASTM B 409 (Ref.5), sheets and plates;
- ASME SB 564 - ASTM B 564 (Ref.6), forgings

Subsection NH of ASME code (Ref.7), dedicated to high temperature refers to those 5 standards for N08810 grade.

1.3. Applications and in-service application

The grade UNS N08810 alloy or rather earlier annealed grades referred as alloy 800 or alloy 800H grade 2 have been supplied and used in various nuclear facilities:

- Reactors Peach Bottom and Fort St. Vrain high temperature, helium-cooled steam generators (Reference 9).
- Reactor THTR 300 (High temperature Thorium Reactor) Schmehausen.

In the case of the THTR reactor, it seems that the specification of the exchange tubes of the superheater is a finer grain size and a more limited carbon content of 0.03 to 0.06% (References 8 and 9) and it would be the grade known as DE by reference 13.

For use in the THTR reactor temperatures limited to 550°C, a grade called Rk (“recrystallised”) , treated at 950-1000°C, was selected (Ref.13): it is not the alloy UNS N08810.

The true 800H grade of Reference 13, with coarser grains is characterized in a temperature range (up to 1000°C) that matches the conditions of use projects VHTR helium and hot loops made in Germany. We ignore a possible use in the THTR. Only KTA 800H grade will be now considered in this document.

The Alloy 800 was used in steam generators of:

- Pressurized water reactors of German design
- SUPERPHENIX sodium fast reactor

For these applications, resistance to corrosion is the primary objective of choice. A large grain size is not favorable, neither to tensile properties, nor resistance to corrosion. For SUPERPHENIX an annealing heat treatment at 980°C, giving a fine grain is specified: it is no longer the alloy UNS N08810.

2. PHYSICAL PROPERTIES

2.1. Young Modulus

Young modulus values are supplied in European Standard (Ref. 1), ASME (Ref. 12) and KTA (Ref. 13) (cf Table 3). Figure 1 compares these three data sources. As the values from the European standard and the Draft KTA are similar and less optimistic than the values from ASME, it is proposed to use current European standard values for ARCHER project.

2.2. Thermal expansion

Mean thermal expansion values are supplied in European Standard (Ref. 1), ASME (Ref. 12) and KTA (Ref. 13) (cf Table 4). Figure 2 compares these three data sources. As the values from the European standard and the Draft KTA are similar and less optimistic than the values from ASME, it is proposed to use current European standard values for ARCHER project.

3. BASIC MECHANICAL CHARACTERISTICS: CONVENTIONAL LIMIT OF ELASTICITY AT 0.2%

The table Y-1 of Section II Part D of ASME (ref.12) is appropriate for S_y up to 538°C (1000 ° F). Additional data are obtained directly from Table I-14.5 of NH 1050 ° F to 1500 ° F (565 to 816°C). But there is obvious disagreement between these two data sources (Figure 3), as already mentioned by reference 14.

Data from Table I-14.5 of the sub-section NH of editions before 2004 to the range 1050 ° F to 1500 ° F (565 to 816°C) give the minimum curve of Figure 4:

- The disagreement between 538 and 565°C is largely absorbed
- Some data from providers confirmed the magnitude of the corrected data ASME.

The minimum yield strength specified in reference 1 are also plotted on figure 4: they are significantly lower than those of ASME when the temperature reaches 100°C.

The results of references 10 and 11 are compared in Figure 5 to those of ASME and of European standard and KTA. The corrected minimum values S_y from the tables of ASME appear overly optimistic. The minimum European standards and KTA values are more representative of experimental data, especially from those of reference 10.

It is proposed to use KTA values because they are available up to 800°C (European standard values are only available up to 550°C).

4. CREEP DATA

4.1. Creep rupture stress Sr

The sub-section NH of the ASME III (ref. 7) provides minimum values of the creep rupture stress versus time at different temperatures from 450 to 750°C.

Reference 13 provides the minimum values of the creep stress for three types of alloy products 800 or 800H: Creep stress values of grade 800H are supplied in the temperature range 700-1000°C. This does not reflect usage in THTR but those projects VHTR helium and hot helium loops made in Germany.

Estimates of the creep rupture stress (mean values) for periods of 10 000, 100, 000 and 200 000 hours at 700°C to 1000°C are given in Reference 1. They are very similar to the mean values of the Reference 13 for 800H grade. It is specified in Reference 1 that mean creep data have been prepared by the European Creep Collaborative Committee (Ref. 15).

Figure 6 gives the comparison of Sr values between these data source from 700°C to 900°C. As minimum values from ASME at 700°C seems too optimistic, it is proposed to use Sr values from KTA (Ref. 13) for ARCHER project.

4.2. Norton's law

Minimum creep rate data are only proposed in reference 13. Results are however in accordance with data from NIMS on 800H plate (internal reference), as shown in figure 7.

Table 7 gives the minimum creep rate as a function of the stress σ and the temperature, and the corresponding Norton's factors k and n, as $d\varepsilon/dt = k\sigma^n$.

5. FATIGUE

Fatigue curves are supplied from 20°C to 900°C by KTA (Ref. 13), and only up to 760°C by Subection NH of reference 7. Both data are in coherence, but as ASME values are limited at 760°C, it is proposed to use the curves from reference 13, given in Figure 8. It should be noticed that the criteria used by ASME and KTA to assess the number of cycles to failure has to be compared, as well as the process of fatigue damage calculation.

6. EFFECT OF HTR ENVIRONMENT

Corrosion of 800H under HTR Helium at high temperature is characterized by:

- The formation of a thick Cr_2O_3 layer with Ti content,
- Internal Al_2O_3 oxidation, mainly at the grain boundaries.

According to the literature, this behavior leads to limit 800H use at temperatures below 850°C. In the reference [13], behavior of Alloy 800H in HTR-Helium is indeed indicated only above 800°C.

7. WELDING COEFFICIENTS

Welding coefficients are provided in reference [13]. Nevertheless these coefficients are not representative of a laser beam process and can not be used for the ARCHER IHX analysis.

8. CONCLUSION

Sum-up of the proposed design data for ARCHER project

- Young Modulus

Temperature (°C)	20	100	200	300	400	500	600	700	800
E (10^3.MPa) NF EN (Ref. 1)	200	190	185	175	170	160	155	145	140

- Mean thermal expansion

Temperature (°C)	100	200	300	400	500	600	700	800
α (10^{-6}/ K) NF EN (Ref. 1)	15,4	16	16,5	16,8	17,2	17,5	17,9	18,3

- Minimum limit of elasticity at 0.2% (MPa)

Temperature (°C)	20	100	300	500	700	800		
Sy KTA (Ref. 13)	170	140	95	80	75	70		

- Minimum stress to rupture (MPa)

°C	Time in hours	100	1 000	3 000	10 000	30 000	100 000	200 000	300 000
700	KTA Ref.13	116	84	72	59	47	37		28,4
800		65	46	38	31	25	19		14,1
900		30	20	17	13	10	7		4,9

- Norton's law parameters

Temperature (°C)	600°C	700°C	750°C	800°C	850°C
K (%/h)	1,11E-24	4,86E-18	1,24E-16	1,88E-14	1,83E-12
n	9,08E+00	8,00E+00	7,42E+00	6,81E+00	6,17E+00

- Fatigue design curve
    - o See Figure 8

FIGURE 1: YOUNG MODULUS

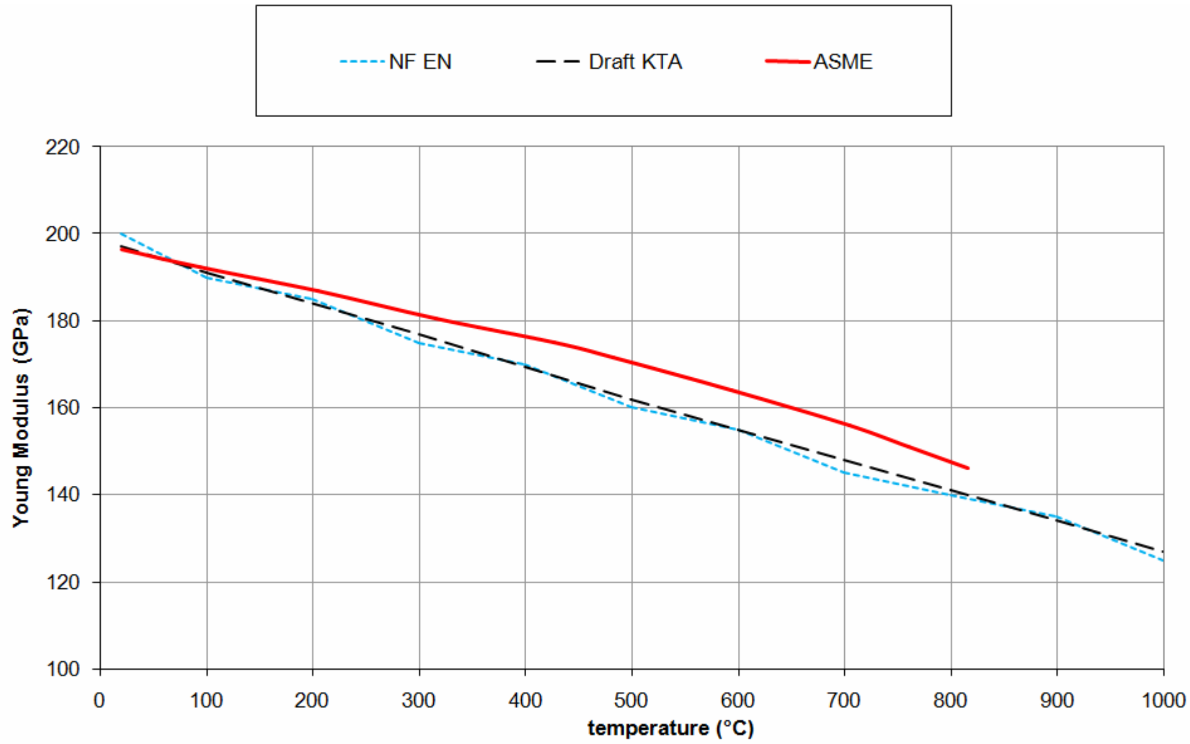


FIGURE 2: THERMAL EXPANSION

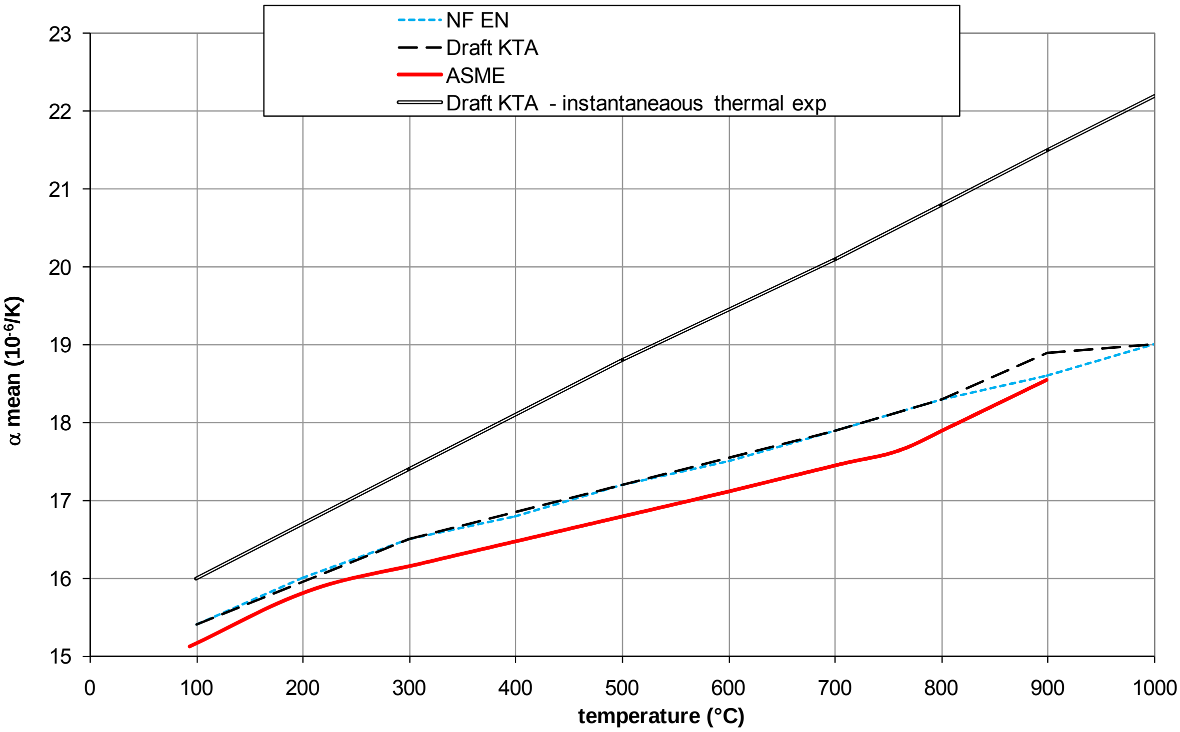


FIGURE 3:
YIELD STRENGTH: MINIMUM VALUES ACCORDING TO ASME II D ET III NH I 14.5 EDITED IN 2010 AND NF EN STANDARD

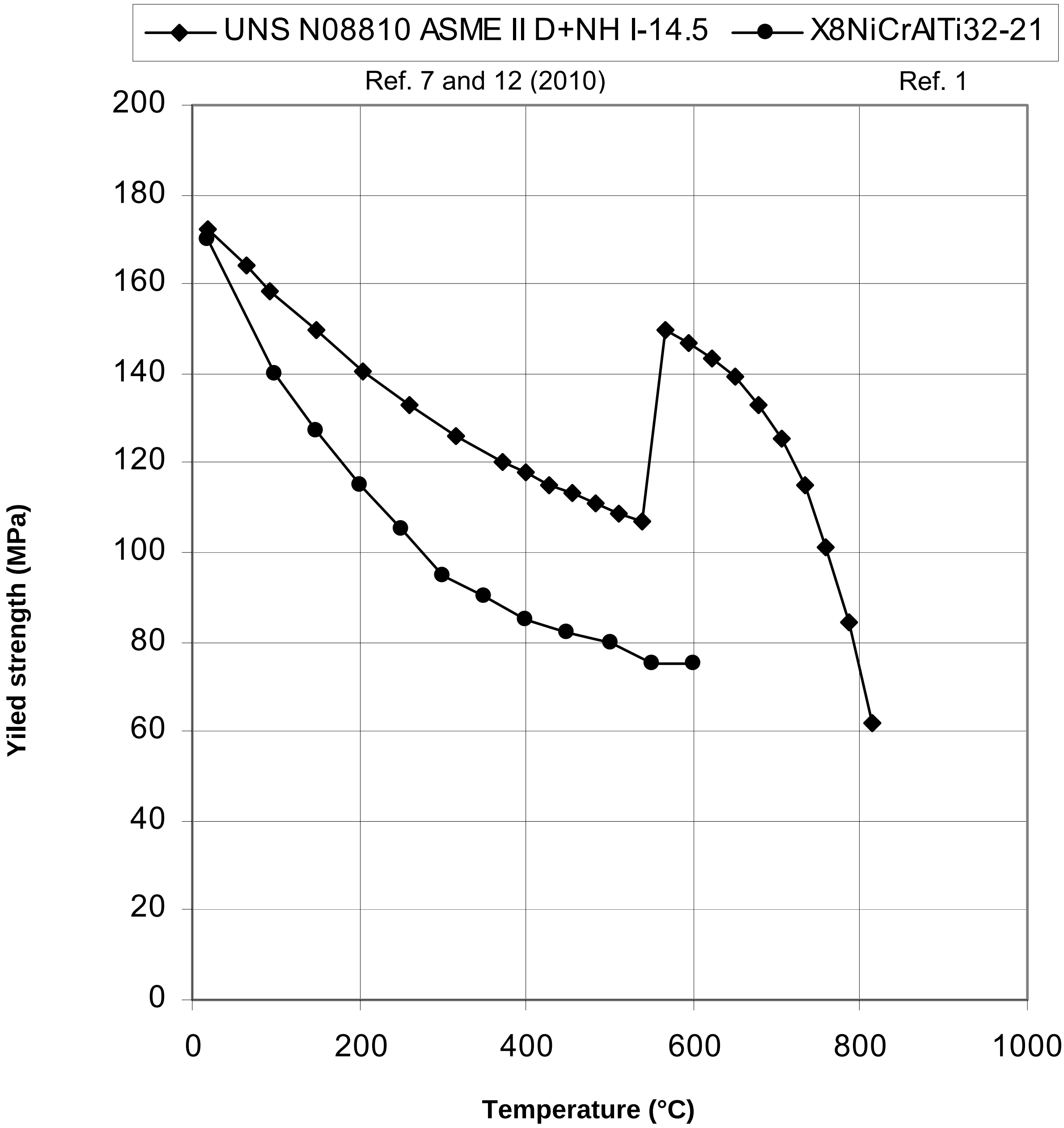


FIGURE 4: YIELD STRENGTH: MINIMUM VALUES (800H DATA FROM ASME-NH EDITED BEFORE 2004) AND NF EN STANDARD

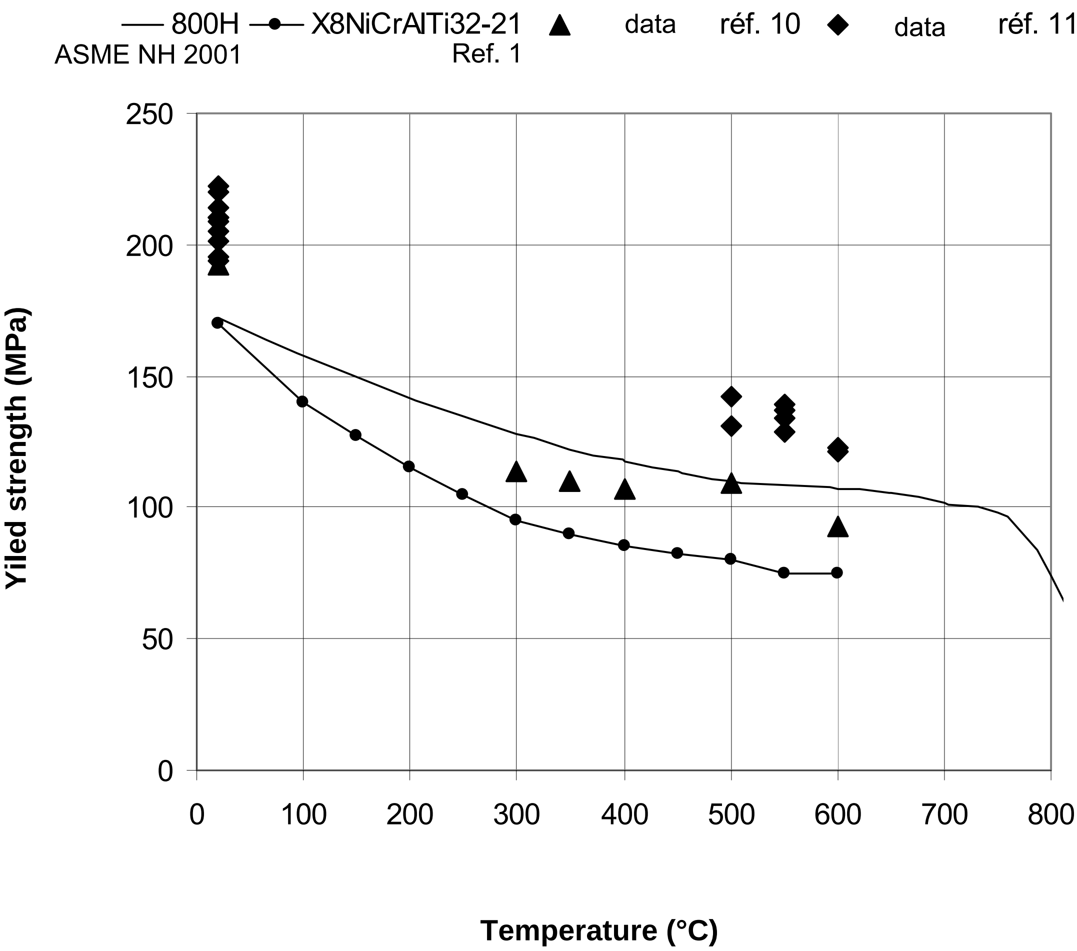


FIGURE 5: YIELD STRENGTH: MINIMUM VALUES (800H DATA FROM ASME-NH EDITED BEFORE 2004 AND EN STANDARD (REF. 1) AND KTA (REF. 13)

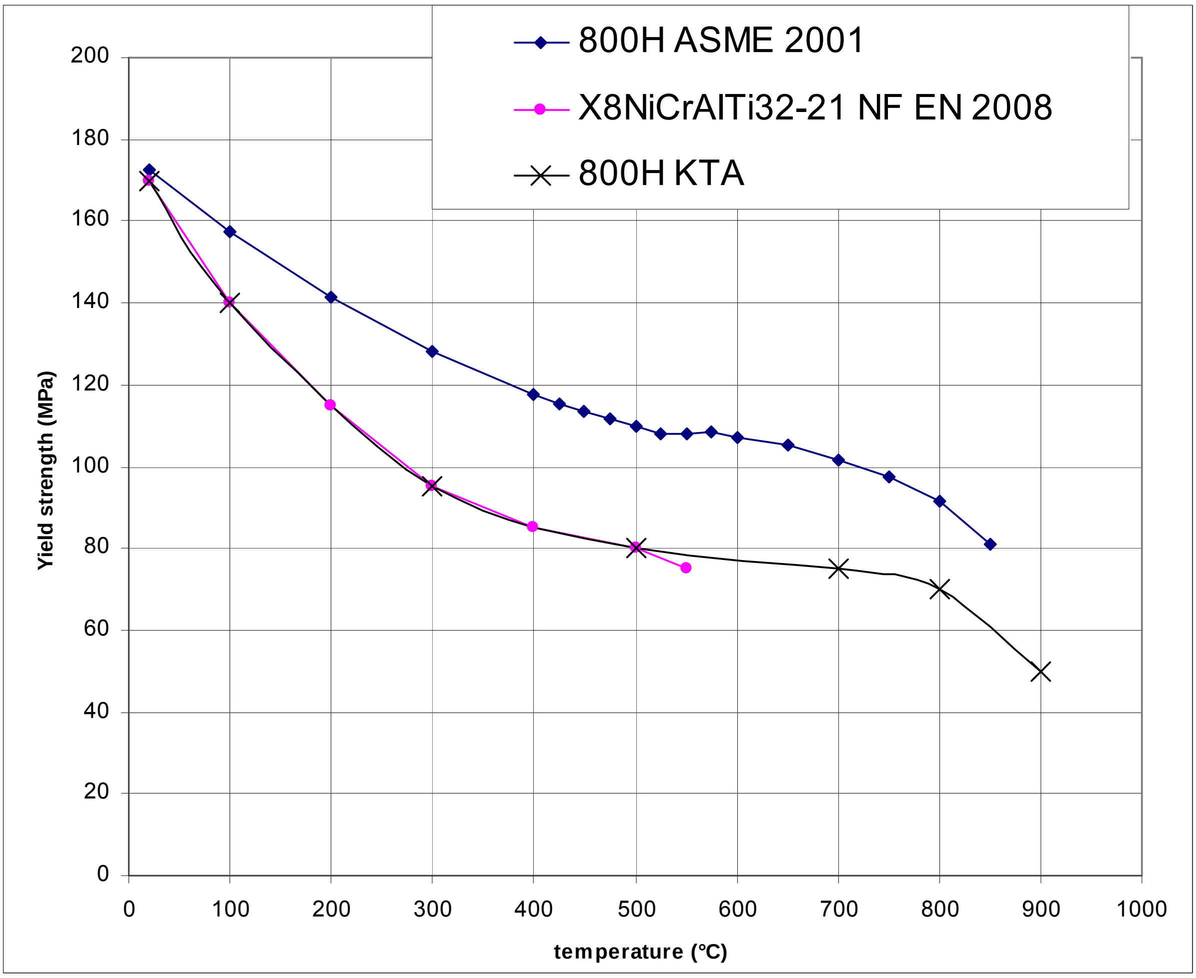


FIGURE 6:
STRESS TO RUPTURE (550 et 650°C)

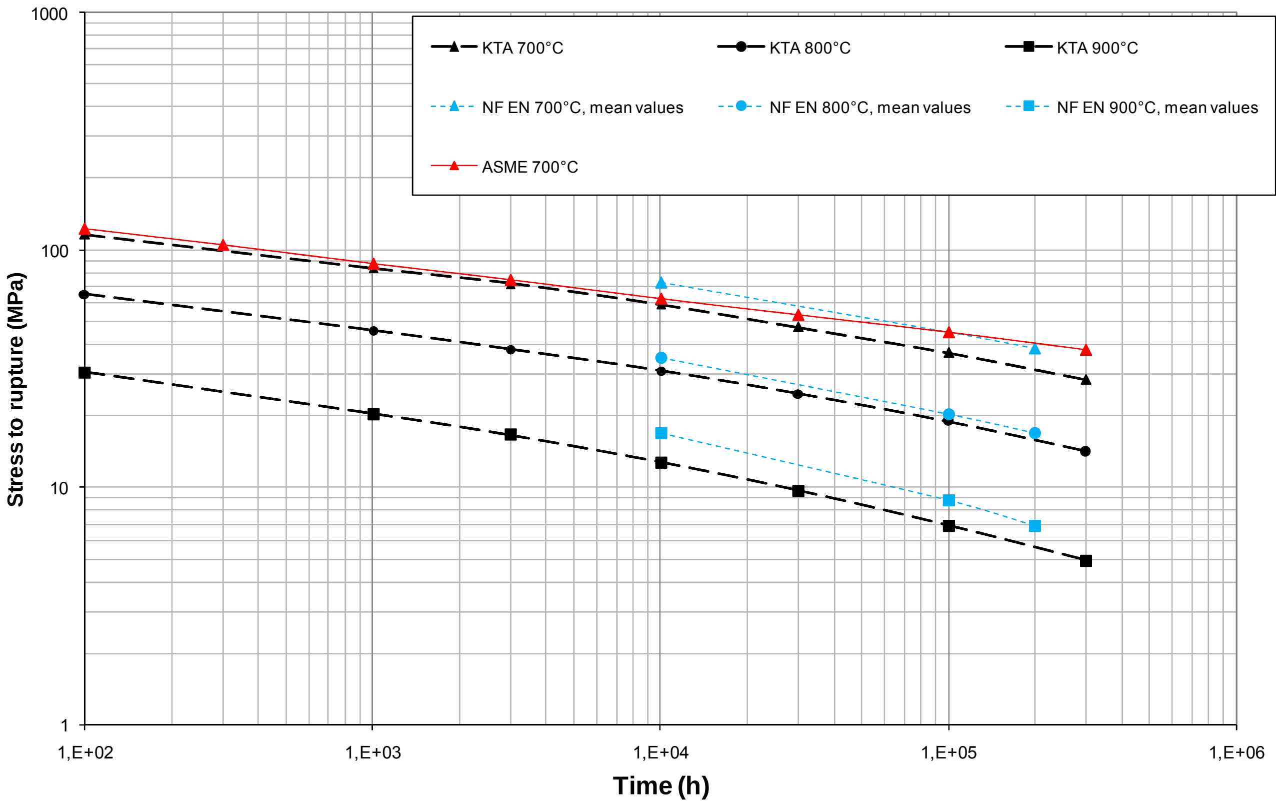
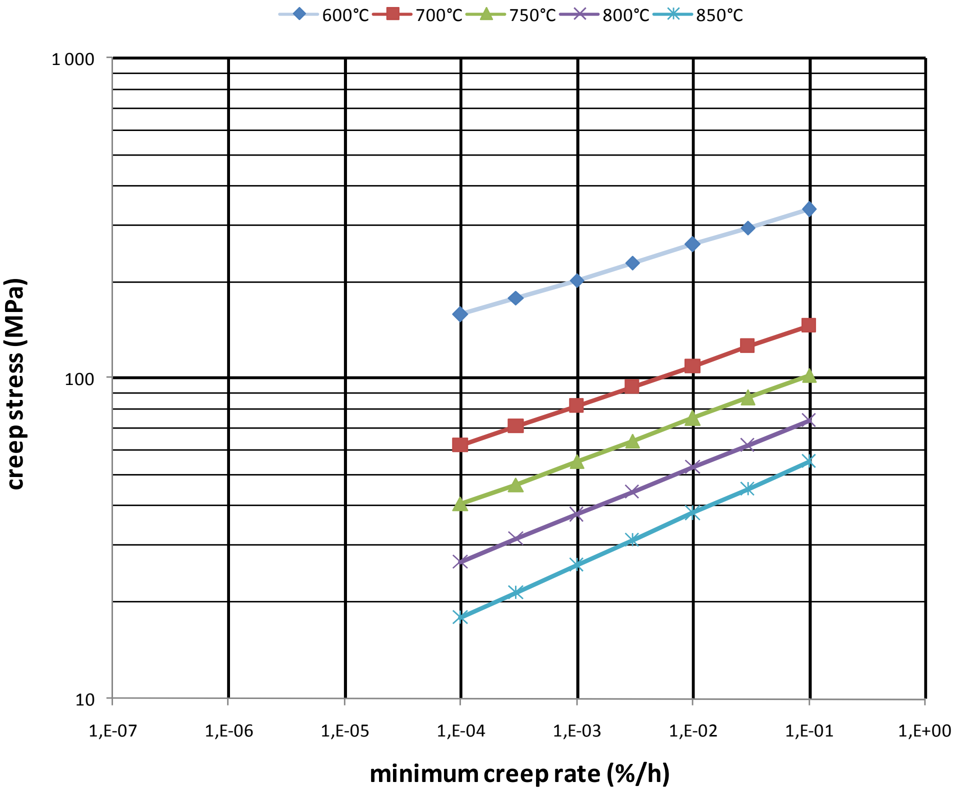


FIGURE 7:
MINIMUM CREEP RATE(REF. 13), COMPARISON WITH NIMS DATA



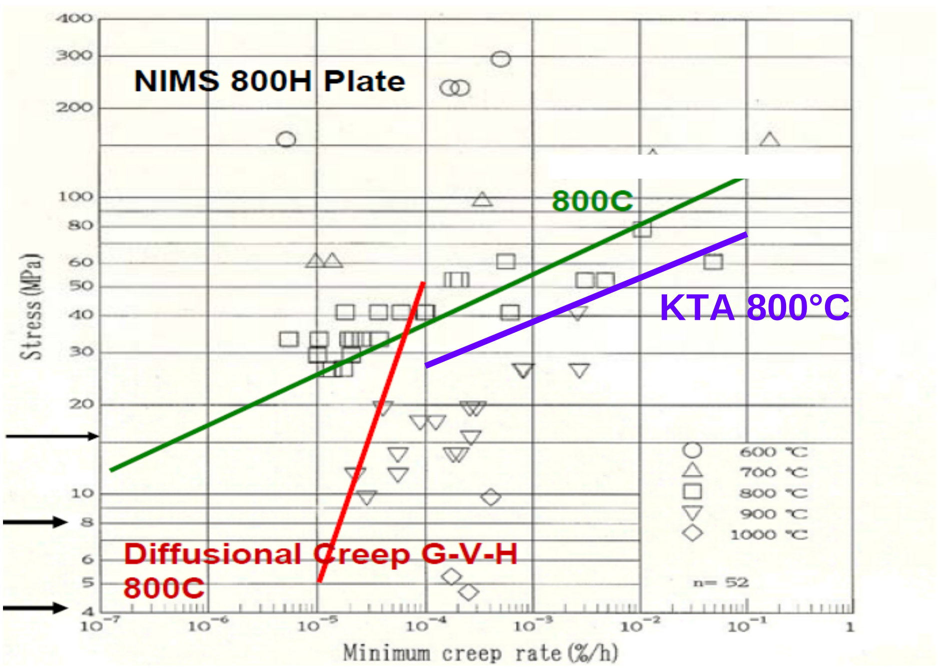


FIGURE 8:
FATIGUE DESIGN CURVES (REF. 13), COMPARISON WITH DESIGN DATA AT 760°C FROM (REF. 7) (★)

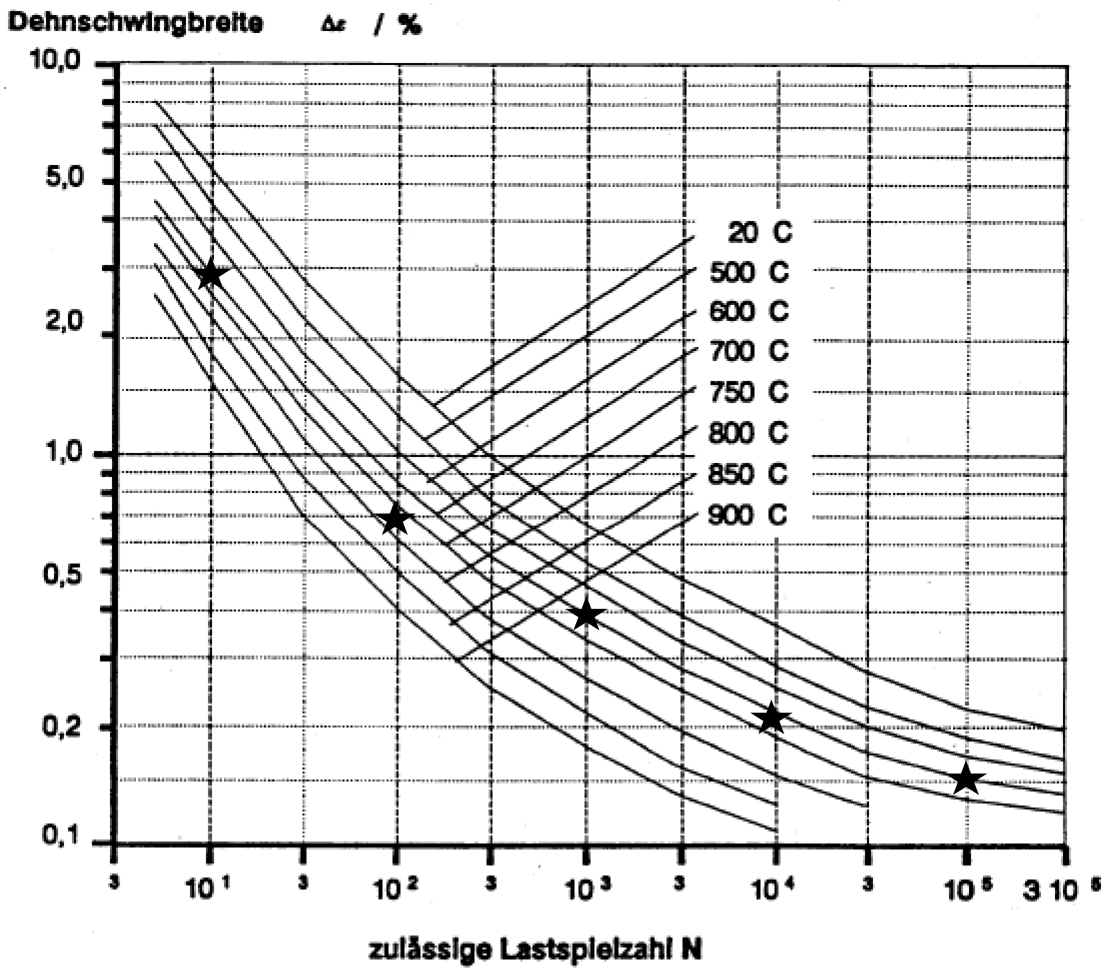


TABLE 1:

CHEMICAL COMPOSITION (WT %)

	Solution annealed condition	C	Si	Mn	P	S	Ni	Cr	Al	Ti	Al + Ti	Cu
X8NiCrAlTi32-21 EN 10302	1100-1200 Air cooled or faster	0,05 - 0,10	≤ 0,70	≤ 1,50	≤ 0,015	≤ 0,010	30,00-34,00	19,00-22,00	0,20-0,65	0,25-0,65	≥ 0,45	≤ 0,50
UNS N08810 ASTM SB 163 ASTM SB 407 ASTM SB 408 ASTM SB 409 ASTM SB 564	>1121°C	0,05 - 0,10	≤ 1,0	≤ 1,50		≤ 0,015	30 -35	19 - 23	0,15-0,60	0,15-0,60	≥ 0,30	≤ 0,75
800H (KTA)	1100-1200°C Air cooled	0,05 - 0,10	≤ 0,70	≤ 1,50	≤ 0,015	≤ 0,010	30,00-34,00	19,00-22,00	0,40 0,75	0,25-0,65		≤ 0,45

TABLE 2:

SPECIFIED MECHANICAL PROPERTIES

STR	Product	Sy (MPa)	Su (MPa)	A %
EN 10302	Long products	≥ 170	500 - 750	≥ 30*
SB 163	seamless tubes for heat exchanger	≥ 172	≥ 448	≥ 30
SB 407	seamless tubes	≥ 170		≥ 30
SB 408	bars and rods	≥ 170	≥ 450	≥ 30
SB 409	sheets and plates	≥ 170		≥ 30
SB 564	forgings	≥ 172	≥ 448	≥ 30
* A % ≥ 35 for flat products				

TABLE 3: YOUNG MODULUS

Temperature (°C)	**20**	**100**	**200**	**300**	**400**	**500**	**600**	**700**	**800**
E (10^3.MPa) **KTA** (Ref. 13)	197,0	191,0		177,0		162,0		148,0	141,0
E (10^3.MPa) NF EN (Ref. 1)	**200**	**190**	**185**	**175**	**170**	**160**	**155**	**145**	**140**
Temperature (°C)	**21**	**93**	**204**	**316**	**427**	**482**	**593**	**704**	**815**
E (10^3.MPa) **ASME** (Ref. 12)	196,5	192,4	186,8	180,6	175,1	171,7	164,1	155,8	146,2

TABLE 4: MEAN THERMAL EXPANSION

Temperature (°C)	**100**	**200**	**300**	**400**	**500**	**600**	**700**	**800**
α (10^{-6}/ K) **KTA** (Ref. 13)	15,4		16.5		17.2		17.9	18.3
α (10^{-6}/ K) NF EN (Ref. 1)	**15,4**	**16**	**16,5**	**16,8**	**17,2**	**17,5**	**17,9**	**18,3**
Temperature (°C)	**93**	**204**	**316**	**427**	**482**	**593**	**704**	**760**
α (10^{-6}/ K) **ASME** (Ref. 12)	15,1	15,8	16,2	16,6	16,7	17,1	17,5	17,6

TABLE 5:

MINIMUM LIMIT OF ELASTICITY AT 0.2% (MPa)

Temperature (°C)	**20**	**100**	**200**	**250**	**300**	**350**	**400**	**450**
S_y ASME (2001)	172	158	141	134	128	122	118	113
Sy KTA (Ref. 13)	**170**	**140**			**95**			
S_y (Ref 1)	175	140	115	105	95	90	85	82
Temperature (°C)	**500**	**550**	**600**	**650**	**700**	**750**	**800**	
S_y ASME (2001)	110	108	107	105	102	98	74	
Sy KTA (Ref. 13)	**80**				**75**		**70**	
S_y (Ref 1)	80	75	(74)	(73)	(71)	(68)	(51)	
( ) : estimated values = ASME values x 0,69 or (75/108)								

TABLE 6:

**CREEP RUPTURE STRESS SR – MINIMUM VALUES (MPa)
COMPARISON BETWEEN STANDARDS AND CODE**

		Time in hours							
°C		**100**	**1 000**	**3 000**	**10 000**	**30 000**	**100 000**	**200 000**	**300 000**
700	ASME (Ref. 7)	123	88	75	62	53	45		38
	KTA Ref.13	**116**	**84**	**72**	**59**	**47**	**37**		**28,4**
	NF EN (Ref.1)				73		44,8	38,2	
800	ASME (Ref. 7)								
	KTA Ref.13	**65**	**46**	**38**	**31**	**25**	**19**		**14,1**
	NF EN (Ref.1)				35,2		20,4	17	
900	ASME (Ref. 7)								
	KTA Ref.13	**30**	**20**	**17**	**13**	**10**	**7**		**4,9**
	NF EN (Ref.1)				16,8		8,76	6,91	

TABLEAU 7:

800H MINUMUM CREEP RATE (REF. 13)

minimum creep rate (%/h)	**600°C**	**700°C**	**750°C**	**800°C**	**850°C**
	Stress (MPa)				
1,00E-04	1,58E+02	6,20E+01	4,07E+01	2,68E+01	1,79E+01
3,00E-04	1,78E+02	7,10E+01	4,67E+01	3,15E+01	2,14E+01
1,00E-03	2,03E+02	8,20E+01	5,50E+01	3,76E+01	2,61E+01
3,00E-03	2,29E+02	9,40E+01	6,40E+01	4,42E+01	3,12E+01
1,00E-02	2,62E+02	1,09E+02	7,50E+01	5,30E+01	3,79E+01
3,00E-02	2,95E+02	1,26E+02	8,70E+01	6,20E+01	4,52E+01
1,00E-01	3,37E+02	1,46E+02	1,02E+02	7,40E+01	5,50E+01
k	1,11E-24	4,86E-18	1,24E-16	1,88E-14	1,83E-12
n	9,08E+00	8,00E+00	7,42E+00	6,81E+00	6,17E+00

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Material Data Review of Thin-Section Alloy 800H for Reactor Heat Exchanger Applications

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Summary

A mock-up intermediate heat exchanger for future (Very) High Temperature Reactors will be tested within the ARCHER EU Project that is running over the period 2011-2015. It is planned to use thin-section Alloy 800H within the heat exchanger, with sheet thicknesses of 1.5mm being achieved by machining from 4mm hot rolled plate. Although a significant amount of material test and property data are available for thicker section Alloy 800H (typically plate or bar material exceeding 15mm section), the amount of testing on Alloy 800H or other variants of Alloy 800 in section of 5mm or less is low. This report summarises the data that has been found, and the theoretical aspects of material behaviour in such thin section, forming part of the deliverable D43.51 of the ARCHER Project "Review of existing data on Alloy 800H (focus on thin sheets)", prior to further testing later in the Project.

It is anticipated that some reduction in stress rupture strength values will occur in thin sections, and depends partly on grain size. However, the present test data in air in the literature with testing times of less than 5kh, broadly suggests that the thin section material has comparable creep properties to thicker section material. Therefore, at least until further testing has taken place within the project and elsewhere, the property values recorded in codes and standards are the most appropriate for the design of the intermediate heat exchanger for the ARCHER WP43 rig tests. A further slight reduction in strength values is possible with the effect of "steam cycle" impure helium, with the effect exacerbated if carburising conditions are likely. It is recommended that both matters be considered in more detail prior to declaring the materials properties of Alloy 800H for the (V)HTR demonstrator, and successor plant.

Introduction

An Intermediate Heat Exchanger (IHX) is required in the (Very) High Temperature Reactor (V)HTR to keep separate the reactor primary coolant circuit, and the secondary circuit fluid intended for a range of industrial purposes. Within the ARCHER EU Project [1], Work Package 4.3 "Intermediate Heat Exchanger" the technology is being further developed from previous initiatives, including testing of a mock-up in the CLAIRE loop of CEA. Heat exchange takes place in a compact cross-flow heat exchanger at low net pressure over thin-section material, with Alloy 800H being selected at a section thickness (after machining) of about 1.5mm. Material properties of such thin section material are being investigated in the ARCHER project, since most of the existing property data has concentrated on thicker section material (typically bar or plate at sections of 15mm or greater).

This present review concentrates on the available literature data on thin section Alloy 800H, and particularly seeks to consider the mechanical properties used for design of the mock-up. As such it forms part of the Deliverable D43.51 of the ARCHER Project "Review of existing data on Alloy 800H (focus on thin sheets)".

Materials and Property Requirements for IHX

The material property design data for Alloy 800H has been summarised in [2], essentially drawing on published design property compilations in particular the ASME Boiler and Pressure Vessel Code [3], and the KTA draft property compilation [4]. That report essentially also defines the property requirements for the IHX, which can be summarised as shown in Table 1.

Table 1: Temperature-Dependent Properties Required for Alloy 800H

Thermophysical Properties	Coefficient of Thermal Expansion Modulus (static/ ~~dynamic~~) ~~Thermal Conductivity~~ ~~Heat Capacity~~ ~~Density~~ ~~Poisson's Ratio~~
Mechanical Properties	Proof Stress - Minimum Ultimate Tensile Strength - Minimum Creep Rupture Strength (as function of time) - Minimum Weld Strength Reduction Factors Minimum creep rate – Mean ~~Creep Strength for strain limit (as function of time)~~ ~~HCF (as function of mean stress)~~ LCF

Note: Properties not needed for the IHX design, and therefore not included in [2] but potentially available if required, are indicated by struck-through text.

Material Production

In general, sheet material in the range of 3mm-5mm section can be produced by two related production methods: hot-rolling and cold-rolling. In both cases, it is likely (though not demanded by the standards) that the material is originally Vacuum Induction Melted (VIM) or Electric Arc melted followed by a secondary Electro-Slag Remelting (ESR), then rough forged or rolled into rectangular slabs.

Hot-rolling of the slab takes place with two or more rolling campaigns with intermediate stress-relief heat treatments. Hot-rolling is quite commonly used to produce sheet of 5-6mm section, and it is sometimes used to produce sheet down to ~3mm section, though cold rolling is increasingly used for sheet below 5mm section, as described below. During the hot-rolling process, the material dynamically recrystallises, that is, it recrystallises as the material is rolled. The skill of the material manufacturer ensures that sufficient deformation is introduced through the section, and the temperature is low enough that recrystallisation occurs uniformly, and so the resulting grain growth size is kept uniform. After the final rolling campaign, the material is given its Quality Heat Treatment according to the requirements in the specification. Often the annealing stage will be performed in a “bright annealing” furnace, that is with a blanket of non-oxidising gas such as nitrogen, argon, or nitrogen + hydrogen (typically “cracked” ammonia).

Depending on future usage, and/or the processing capabilities of the manufacturer, the hot-rolling process may be carried out on a strip mill either continuous or reversing between coilers. Additionally, the final anneal can be carried out in a strip anneal furnace (that is, the strip is passed through a furnace continuously between coilers). Typically, the manufacturer will ensure that the final pass of the strip through the rolling mills will introduce sufficient work so that further recrystallisation will occur during the Quality Heat Treatment, otherwise excessive grain growth could occur. Even so, it is quite typical that hot-rolled material has slightly larger grain size at the centre of the section (where deformation is less, and heat retention during rolling greatest) and towards the centre of the sheet in the long-transverse orientation (deformation generally being greater towards the edge of the rolls). Hot-rolling is considered to be more difficult to achieve tight tolerances in the finished section, and consequently the reduction in the final roll before the Quality Heat Treatment is more variable, therefore with the potential to introduce greater variability in the finished grain size.

Cold rolling is used as a final stage (can be two or more rolling campaigns, with intermediate stress relieving treatment) on previously hot-rolled material. Cold rolling has the advantage of giving closer tolerances on the final section, and also no dynamic recrystallisation takes place. Instead, recrystallisation occurs during the Quality Heat Treatment annealing stage, driven by the stored cold work. The final grain size is often more uniform in cold rolled product, since it is easier to introduce a more uniform degree of work through the section. Moreover, it can be kept small, or permitted to grow depending on the parameters of the annealing process. Almost without exception, bright annealing is used to provide a good surface finish, particularly on coiled strip.

In both hot and cold strip rolling it is not possible to use cross-rolling, as can be applied in single sheet manufacture. The starting forged ESR billet can be cross-rolled to increase width, though this will have little effect on inclusion defined microstructural directionality. Modern ESR produced rolling feedstock has good homogeneity and usually very low inclusion content compared to other air melted/cast material. Preferred orientation (texturing) is reduced by proper process control in both hot and cold rolled stages.

In summary of this section, Alloy 800H sheet with a thickness of 3-5mm of typical interest for the IHX can be produced by a variety of melting practices (VIM or electric arc primary melting, often followed by secondary electroslag remelting) and rolling routes. When compared to finishing by cold rolling, hot rolling has the potential to generate more variation grain size between products produced by the same route. It may also result in slightly larger grain size within the centre plane of the plate.

Component Manufacture

A number of processes can occur during the manufacture of a component that can alter its properties in service. These include, but are not limited to:

- Cold bending, stamping and/or other cold forming operations. The cold work alters the tensile properties, and so it may be necessary to stress-relieve worked areas. Retained levels of cold work may recrystallise, as summarised in [5], Figure 1.

- Welding or brazing. The input of heat during the process, and the subsequent heat treatment procedure has the potential to cause grain-growth. Such heating causes expansion and contraction and hence introduces residual stresses. Both effects can lead to different precipitation response.

- Machining. Excessive machining rates can inadvertently lead to work at the surface and hence residual stresses and the possibility of local recrystallisation.

With reference to the IHX, austenitic steels in thin sheet form have already been used to make compact cross-flow heat exchangers for lower temperature use (up to about 550°C), and it is not thought that the manufacture of the IHX using Alloy 800H will be substantially different. Nevertheless, care should be taken to ensure that the cold work introduced through stamping is kept below the onset of recrystallisation line shown in Figure 1.

Weldment Relaxation Cracking

Weldment relaxation cracking has occurred in several austenitic alloys, and is known by a variety of alternative names, including strain-age cracking, stress-induced cracking and reheat cracking (see [6] and [7] for a review). It seems particularly prevalent in the process industries where sections are typically larger than Alloy 800H usage in nuclear plant. According to those authors, the common characteristics are typically:

- Service temperatures of 500 to 750°C
- Cracking during fabrication heat treatment or after only brief service (less than 2 years)
- Cracking within one year of repair or modification
- Intergranular cracks, often associated with voids
- Brittle crack behavior with no associated deformation
- Cracks located in the heat affected zones adjacent to welds or in heavily cold deformed areas
- Presence of a nickel-rich metallic filament on the cracked grain boundary often enclosed by a chromium-rich oxide layer

Van Wortel in [6] attributes susceptibility to relaxation cracking to strains induced by welding, or cold deformation or cyclic loading, leading to an enhanced dislocation density, and resulting in a rapid precipitation of fine $M_{23}C_6$ type intragranular precipitation limiting further deformation. Cracking is said to result after relaxation strains of less than 0.2%. Generally, there is a zone denuded of precipitates adjacent to the grain boundaries, and cracking always takes place on the grain boundary, usually from the free surface and is often characterised by a nickel-rich metallic filament. Vickers hardness in the weldmetal and immediately adjacent is often in excess of Hv_5=200. Attempts to control weldment relaxation cracking have included annealing Alloy 800H at 980°C for 3h prior to welding or repair welding. A post weld heat treatment cycle of 980°C 3h, is also recommended if Type 617 consumables have been used, although 875°C for 3 hours for Ni base type 82 or Fe-base consumables is said to suffice [8]. As yet, though, there is no fully industry-accepted approach to the matter.

The lack of a fully accepted approach to reduce weldment relaxation cracking is perhaps partly because the investigative work of Van Wortel generally fails to acknowledge any consequence of γ' precipitation that is prevalent at similar temperatures in many of the alloys studied, and in particular Alloy 800H and

IN617. Moreover, Alloy 800H can suffer "sensitisation" at grain boundaries due to coarse $M_{23}C_6$ precipitation leaving a chromium-depleted zone alongside the boundary when aged at 500-700°C. Close to the surface, the $M_{23}C_6$ oxidises and the depleted layer also suffers attack, leaving a nickel-rich filament. Annealing at temperatures of 900°C and above can redistribute the chromium content, thereby reducing the potential for "sensitisation"; it will also take the γ' precipitation back into solution.

In the IHX, the laser welding process is of course quite different to the Manual Metal Arc (MMA) welding of the process industry. The IHX has much smaller welds, less heat input and hence less potential for the high residual stresses sometimes seen in MMA weldments. Furthermore, precipitation hardening alloys, known to be susceptible to relaxation cracking have already been fabricated into similar plate heat exchangers without relaxation cracking being evident (though, with application at lower temperatures). On the other hand, the IHX weldments are on thin-section material with a relatively large grain size (see Discussion section) that will not be stress-relieved after welding. At the very least, it will be advisable to perform micro-hardness traverses of the weldment to look for hardness values approaching or exceeding Hv_5=200.

Metal Dusting

The term metal dusting refers to attack of steels and nickel base alloys in highly carburising environments, generally with the carbon activity, $a_C > 1$, and occurs commonly at intermediate to high temperatures (400-900°C) in process plant, syngas environments, and heat treatment facilities where such environments can exist. Local attack by the carburising atmosphere through defects in the normally protective chromia scale are thought to result in M_3C carbides which eventually lead to graphite formation on the surface (see Section 5.3 of reference [6] for further details and references). Metal dusting is well known to occur in Alloy 800H, see [10]. Metallurgical approaches to reducing carburisation and the associated metal dusting include ensuring that chromium is not depleted in surface layers (that may occur eg. by electrochemical polishing), and specifying a smaller grain size to facilitate faster chromium diffusion to the surface to form a better chromium oxide scale [6]. In general, the highly carburising conditions necessary for metal dusting are not seen in HTR environments (the carbon activity is generally below 1) and so can generally be disregarded. However, some carburisation may be possible in gas turbine cycles, as described below.

Exposure in Impure Helium

Much of the work in the 1980s and 1990s on the effect of impure helium on the surface and mechanical properties has been reviewed by Natesan et al [11], with later additional experimental work reported by Natesan et al in 2009 [12] primarily related to Pebble Bed Modular Reactor. Helium is impure in the reactor circuit for several reasons, not least the presence of carbon from the moderator and low levels of water. Contamination by H_2, H_20, CH_4, O_2 and CO_2 must be considered, and the actual composition of the gas depends on several factors, primarily the temperature and pressure of gas, and the presence of solid phase materials such as graphite and metal/oxide/carbide surfaces.

Typical gas compositions used in the experiments can be judged from Table 2 (Table 3.2 from Natesan's review). The resistance to attack by impure helium in Alloy 800H appears to depend on several factors, which was summarised by the following (citing the work referred to in Table 2):

- The level of impurities varies depending on whether the reactor is based on the steam cycle or gas turbine cycle.
- Thermodynamically, the reactor coolants are reducing to iron and other common metallic oxides.
- Carburisation can be apparent in both steam-cycle and gas turbine cycle compositions, but the level is higher in gas turbine cycle (low oxidation potential) gases.
- Control of oxidation can be effective in reducing carburisation in reactor alloys, but must be reviewed in the light of any effect on graphite oxidation.
- A low CH_4/H_20 ratio tends to stabilise the Cr-based surface oxide, whereas a high CH_4/H_20 ratio can lead to a surface based Cr-carbide and continuous carburisation

- High levels of silicon (~0.5wt% Si) can lead to spallation of the Cr-based oxide. (Though such levels are not uncommon in Alloy 800H.)
- According to Schubert, 1984, a spinel and carbide scale can form with a depth of ~20µm below the surface of Alloy 800H after exposure in reactor gas at 850°C for 10,000h; the maximum metal affected extending approximately 40µm below the surface.

Table 2: Gas chemistries used in investigations of impure helium after Ref. [11]

Table 3.2 Gas chemistries used in various investigations

Gas Identification	Gas composition (in Pa) H_2	CO	CH_4	CO_2	H_2O	N_2	Reference
A	150	45	5	<0.05	5	-	Johnson and Lai 1981
B	20	10	2	0.5	5	-	Johnson and Lai 1981
C	50	5	5	<0.05	<0.05	-	Johnson and Lai 1981
D	50	5	5	<0.05	0.15	-	Johnson and Lai 1981, Brenner and Nilsen 1978
HTR	50	5	2	-	0.1	-	Graham et al. 1981
Decarb.	50	0.5	0.5	-	0.1	-	Graham 1990
Carb.	50	0.15	0.2	-	0.01	-	Graham 1990
JAERI-B	20	10	0.5	0.2	0.1	<0.5	Kondo 1981
ERANS#2	30	10	0.4	0.1	0.3	<0.5	Kondo 1981
PNP	50	2	2	0.1	<0.2	0.5	Kondo 1981, Quadakkers and Schuster 1984, Bates 1984
GE	40	4	2	0.02	0.2	0.6	McKee and Frank 1981
Lab I	150	45	5	-	5	-	Cappellaere et al. 1984
AIDA	150	45	5	-	5	-	Cappellaere et al. 1984
Lab II	50	1.5	2	-	0.15	0.5	Huchtemann 1989

ERANS: Engineering Research Association of Nuclear Steelmaking, Japan.
JAERI: Japanese Atomic Energy Research Institute.
HTR: High Temperature Reactor.
PNP: Prototype Nuclear Process.
GE: General Electric Company.
Lab I and Lab II: Environments in laboratory tests.

Natesan's review similarly considered the effect of impure helium (generally typical of the steam-cycle) which were shown to have a negligible effect on properties such as creep up to 760°C and 10,000h. The work reported is contained in Betteridge, 1978, which is primarily on relatively thick section products (ie. tubes of >5-6mm thickness). Of course, if (after Schubert, 1984, at 850°C) we suppose that the maximum metal affected has a depth of ~50µm, then since this occurs on both sides of a thin sheet it corresponds to 6.6% loss of section of a 1.5mm thickness sheet, but only 1.6% loss of section of a 6mm wall thickness tube. Even so, the effect of thickness loss is likely to be fairly minimal on mechanical properties, particularly as the loss will be less at the lower temperatures envisaged for the IHX (although the durations are considerably longer). Further studies characterising the maximum metal affected and therefore loss in load bearing section, over the temperatures and durations relevant to the IHX may be required.

Relatively few experiments have been performed under cyclic conditions that might be thought likely to promote spallation. As thermal transients are to be deliberately introduced in the CEA rig tests within WP4.3, this might provide a source of material that could be examined metallographically for any evidence of spallation and section loss. On the other hand, as the tests are in air, the rig conditions are far from typical of use within the reactor.

Microstructural Evolution

The earliest work on summarising the precipitation sequences within Alloy 800H seems to have been undertaken by Orr in 1978 [13]. In that study, a lower temperature (circa 650°C) precipitation of γ' phase (Ni_3,[Al,Ti]) was said to occur, with metal (mostly chromium) carbide $M_{23}C_6$ phase dominating at higher temperatures. Primary titanium carbonitride, Ti(C,N), is usually present from the melt.

Spiradek et al [14] in their microstructural investigation found that, after solution treatment, Alloy 800H contains primary Ti(C,N) together with some very thin $M_{23}C_6$ phase at the grain boundaries. Referring to the time-temperature-precipitation diagram in Fig 4 they state that γ' generally occurs below 700°C, and hence was of less consequence for their investigations at 800°C. Generally, fine $M_{23}C_6$ phase was found to precipitate on dislocations generated during creep, or present due to prior cold work, preventing the sub-grain migration. The lowest creep rates at 800°C were evident in the highest predeformed Alloy 800H (cold work of 4.7%), somewhat supporting the theory expounded by Van Wortel, described in the Weldment Relaxation Cracking section of this report. Spiradek et al give direct evidence from TEM imaging of the fine $M_{23}C_6$ phase decorating the dislocation structure, proposing that creep strength is lost as the fine $M_{23}C_6$ phase coarsens with time and/or deformation. Interestingly, in other alloys Ti(CN) is known to precipitate most rapidly, though in Alloy 800H it seems that chromium and carbon combine in $M_{23}C_6$ whilst at temperatures below about 800°C the titanium primarily precipitates, with aluminium in the γ' phase.

Other authors have studied the consequences of long-term ageing of Alloy 800H, such as Konosu et al [15], who examined and creep tested material that had been exposed in service at approximately 650°C for 72,000h. Although that exposure was said to be at low stress (as an inner sleeve to an expansion bellow of a chemical plant), the exposed material had a relatively high hardness of Hv_5 = 185-205, which was attributed to the formation of the γ' phase Ni_3,[Al,Ti], though without direct evidence. The hardness could be reduced by heat treatment with most of the hardening effect removed at temperatures of 900°C and above. The authors also applied a 1100°C solution treatment to some material, and in the as-exposed, 900°C anneal and 1100°C solution treatment condition performed a series of tests at 650°C, attributing the differences in properties according the changes in microstructure. For example, the highest secondary creep rate was observed in the material given the 900°C anneal, since less hardening was to be expected in those specimens. To some extent, this work indicates that the presence of the γ' phase Ni_3,[Al,Ti] may be important, where temperatures are close to its peak of formation, 650°C.

For the IHX in the ARCHER project, the temperature are generally above the peak of formation of the γ' phase Ni_3,[Al,Ti], but nevertheless it will still be expected to form, together with a both Ti(C,N), and $M_{23}C_6$ phases forming, accentuated by prior deformation (eg stamping), strains due to welding, and any creep and fatigue loading applied in service. The consequence of such hardening, and possible associated loss of ductility, may need to be studied further, and much useful information may be learnt from microstructural examination (eg. micro-hardness traverses, optical and electron microscopy) of the specimens tested within the WP43.

Review of Key Material Properties

The following review seeks to find and review any evidence for the effect of thin-section on the mechanical properties of Alloy 800H, and addressing any need for a reduction in strength values relative to thicker section materials, both for the rig design and more generally for future use of such thin sections in the IHX.

Tensile Behaviour

To date, no test data on the effects of thin-section on tensile behaviour in Alloy 800H or similar alloys has been found in the literature, therefore the values from the draft KTA document [4] already included [2], seem entirely appropriate for the design of the heat exchanger.

Fatigue Behaviour

At present a number of references have been found generally relating the fatigue behaviour of Alloy 800H, see references [17] to [20]. However, none of these references contain useful information on the effect of thin-section testing (though fully-reversed loading of thin-section samples is nigh-impossible, due to buckling). Moreover, the present references contain little information on the effect of helium environment. For the time being, and until further references can be found, or new data generated, the values from the draft KTA document [4] already included [2], are appropriate for the design of the heat exchanger.

Creep Rupture Behaviour

A detailed consideration is taking place to extend the properties of Alloy 800H to temperatures from 760° to 850°C in the ASME Section III Subsection NH, see references [25] to [27].

The effect of thin section on creep rupture properties in impure helium is found only rarely in the literature, and often in experiments that do not permit the effects of testing thinner sections in air to be deduced separately from the effects of environment. Ideally, each heat of material would be tested: in thick section form in air; in thin-section form required by the IHX in air; and in the same thin-section form in impure helium. It is notable, that there is little thin-section data available within the JRC-Petten MatDB database, for example (unless it is only available with restricted access). Background Alstom work on Alloy 800 and its varieties have failed to uncover any work on thin-section materials, references [28] to [31]. The European Creep Collaborative Committee has proposed new values for Alloy 800H, reference [32], but again based on data with no thin-section material.

Nevertheless, creep testing has been reported by Bhattacharyya, 1982 [16] on Alloy 800H in air, as part of work to evaluate the properties of thin-section materials for Stirling engines. In that work, the composition of the test materials was reported, but not the production route[1] or the final section size (said to be in the range of 0.79 to 0.99mm). The Alloy 800H was supplied to AMS 5871D (1972). Heat treatment was performed at 1149° in helium for a time of 1h/inch minimum followed by rapid cool or quench, with a final average grain size by the mean linear intercept method found to be 64µm, and an average Rockwell A hardness of 40.0 (range 37.4-42.5). Creep testing was performed in air to ASTM E139, using flat specimens according to ASTM E8, with 6.35mm width and a length/width ratio of 4. A table of data is copied into Table 3 of this report, and the data plotted in conventional rupture plots, and as Monkman-Grant type plots in Figures 2 and 3, respectively. In Figure 2, the KTA draft mean design properties [4] at specified durations have been interpolated with respect to temperature, and plotted on the same figures.

While Bhattacharyya's report does not contain direct comparison between thick and thin-section material of the same heat, and is relatively short-term (<10kh), it is clear to see from Figure 2 that thin-section material when tested in air compares favourably with the KTA draft mean design properties from [4], at least in the temperature range to 815°C. Perhaps at higher temperatures there is more evidence that the longer term test points are beginning to fall below the mean line. Generally, the rupture elongation values on exceed 15% (on 4D), which is acceptable. Though, as might be expected, there is some indication that they will reduce further at longer test durations (>10kh). In Figure 3, A Monkman-Grant type plot of rupture life and minimum creep rate is presented, showing a generally good correlation. Further analysis of the data is presented in Bhattacharyya's original report.

To date, no directly relevant references on thin-section Alloy 800H tested in impure helium representative of potential use in the IHX have been found. As stated above, exposure in impure helium essentially causes either oxidation (steam-cycle impure helium) or carburisation (gas turbine cycle impure helium). The effect in the(V)HTR is likely to be similar to previous work on steam-cycle impure helium, where relatively stable surface oxides or spinel were formed, though that may spall due to higher levels of silicon, but where the principle effect is one of loss in section. In preliminary work on Alloy 800H by Natesan and Shankar [12] in impure, "carburising" helium containing 675 vppm CH_4, a slight reduction (relative to pure helium) in creep life and a lower fracture strain was evident on round samples of 3.8mm diameter, with failure occurring by cleavage at 750°C. At higher temperatures the effect of methane was less, and the failure mode was mixed intergranular / transgranular. Such work should be regarded as cautionary rather than directly relevant to the IHX, since such "carburising" helium is unlikely to be found in either the rig or future steam / process-heat based cycles.

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[1] It is considered most likely that the sheet material tested by Bhattacharyya was cold rolled.

Discussion

It is well-known that thin-section material can have different mechanical properties compared to thicker section material. The main reasons cited are: a direct change in plate manufacturing route or an inadvertent affect of subsequent processing; the effect of having only a small number of grains across the section; and the consequences of an interaction of the surface with the environment (impure helium gas).

Taking these in turn, it is known that the central region of the IHX plate will be 1.5mm thick, having been machined down from 4mm hot-rolled plate in fully heat treated condition. Subsequent stamping operations to introduce the ridges, and welding at the plate edges and to introduce the pipes linking the IHX channels may also need consideration. It is likely that the 3mm section plate will have comparable mechanical properties to other thick section Alloy 800H heats. Provided the machining and stamping operations do not introduce cold work exceeding about 20% or, if they do, a stress relieving treatment is applied, it is unlikely that recrystallisation will result. Furthermore, any potential loss in properties associated with, for example, bright annealing (associated with a chemical reduction of surface oxygen but also, potentially, decarburisation at the surface), will be removed by the machining away of the surface layers.

Turning now to the ratio of grain size to section, early work was performed on other wrought and cast nickel based alloys, but no references have yet been found on Alloy 800H. Richards, 1968 [21] noted that sometimes nickel-based blade forgings (with compositions approximating to Nimonic 90) have higher creep rupture strengths when given a lower temperature solution treatment, resulting in a lower grain size, compared to the conventional wisdom that better creep rupture properties result with larger grain sizes. In a series of experiments on both wrought and cast materials it was found that both a reduction in specimen size (diameter) and increasing grain size (through higher solution treatment temperature) could lead to a reduction in creep life, Fig 5. Gibbons, 1981 [22] performed similar work on Nimonic 90 sheet specimens (0.15 – 1.7 mm thickness) with different grain sizes (40-70µm). Reducing specimen thickness was associated with increasing minimum creep rate and reduced rupture life (up to a factor of x4 in 100MPa tests at 750°C), Fig 6. It was deduced that the creep mechanism was similar in both thin and thick sections of Nimonic 90; but suggested that cavitation of individual grains would be more significant in thinner sections, with thin-section also causing a possible reduction in geometrical constraint, and hence leading to higher grain boundary sliding and higher creep rates.

A specimen geometry effect was also found by Pandey et al 1989 [23], who in tests on Alloy X750 (mean linear grain size of 160µm) in flat, cylindrical and tubular form found that the key parameter defining behaviour was the ratio of area to perimeter, Fig 7. In that work, it was further discussed how the reported properties of thinner-section samples are adversely affected by necking, and by growth of a single dominant crack. Baldan, 1999 [24] considered these approaches in work on conventionally cast MAR M002 with grain sizes (MLI) ranging from 500 to 1000µm and specimens with diameters ranging from 1.1 to 4mm. It was interpreted that the effects of section and grain size could be associated with an increase in both grain boundary sliding in secondary creep, with increased creep crack growth rate in the tertiary regime.

Any interaction of the metal surface with the environment will have a greater effect on mechanical properties of thin section materials compared with the more usual section sizes used in high temperature plant. As has already been presented in the section entitled “Exposure in Impure Helium” the key parameter seems to be the CH_4/H_20 ratio, with higher ratios leading to increased carburisation and lower ratios stabilising the Cr-based surface oxide. Not surprisingly, work on sections typical of steam generator tubes has found little effect of impure helium on creep below 800°C. Nevertheless on 1.5mm sections exposed for several years, the amount of affected metal at the surface will probably have an effect on creep and other properties.

In summary, it seems clear that there is potential adverse effect of both increasing grain size and reduced section on the properties of wrought nickel based alloys, with rupture lifetimes affected by a maximum of about x5 (tests in air on comparable alloys). From the mechanisms proposed to explain such effects, it is likely that they will be similarly evident in Alloy 800H. However, the results of Bhattacharyya on Alloy 800H in air [16] do not presently support the need for a substantial reduction in the design properties. The lack, especially, of creep test data found to date on thin-section Alloy 800H in "steam cycle" impure helium introduces an element of risk in future designs that should be considered, but unless, for some reason, excessive carburisation is suspected is not likely to be of major significance at temperatures below 800°C.

Conclusions

In a preliminary study of the potential effect of thin-section on the properties of Alloy 800H for use in the ARCHER EU Project [1], Work Package 4.3 "Intermediate Heat Exchanger" rig design, it is found that:

1. The thinnest section (1.5mm) IHX plate is being machined from 4mm hot rolled plate. Therefore, the design properties are unlikely to be affected by plate manufacturing route, though there remains a slight risk of greater variability in grain size, and surface area to volume ratio effects.

2. Cold working operations, including stamping, should be kept to below 20% and/or stress-relieved to avoid recrystallisation. Otherwise, there is no evidence to suggest that manufacturing operations currently used to prepare plate heat exchangers in austenitic steels should not be applied to Alloy 800H.

3. Weldment Relaxation Cracking and Metal Dusting, known to affect the use of Alloy 800H in process plant, are unlikely to be significant in the IHX, since laser beam welding will be employed in manufacture, and carbon activity levels will be kept low, respectively.

4. Impure helium is known to degrade properties, but is worse in "carburising" impure helium typical of "gas turbine" cycles rather than the "oxidising" impure helium expected in a "steam cycle". Further work for use of Alloy 800H in (V)HTRs is required to identify the potential for spallation and loss of section, particularly if it is thought that the impure helium could be slightly carburising.

5. No tensile, no fatigue and little creep data have so far been found that are directly relevant to the conditions of the IHX. Creep tests of Alloy 800H in air suggest that there is little effect on creep properties per se of thin-section at temperatures below 800°C.

6. Theoretical approaches, and testing of the effect of grain size and section on creep properties of cast and wrought nickel alloys, suggest that both a reduction in rupture life and an increase in minimum creep rate can occur. In testing, a reduction in rupture life by a factor of up to x4 was evident in extreme cases (approximately corresponding to a 15% reduction in strength).

Presently, therefore, and with some restrictions, there is no evidence suggesting a need to reduce the design properties of Alloy 800H for the design of the ARCHER WP43 rig. On the other hand, the slight but significant effects of both thin-section and the effect of impure helium need further evidence from the literature, or further testing, to determine any strength reduction required for the design of a compact cross-flow IHX in Alloy 800H for the (V)HTR demonstration and successor plant.

Acknowledgements

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Table 3: Stress rupture test data on Alloy 800H thin sheet (0.79 to 0.99mm). (Extracted from Bhattacharyya, 1982)

Test Temp, °C	Applied Initial Stress, MPa	Rupture Life, h	Min Creep Rate, s^{-1}	Total Elongation (4D), %	Duration of secondary creep, h	Time to reach 1% creep, h	Time to reach tertiary creep, h	Effective life, h. Rupture life - time to tertiary
650	276	54.8	2.64E-07	32.2	16	3	19	35.8
650	248	101.4	1.39E-07	26.3	20	5	25	76.4
650	207	309	6.94E-08	15	110	0.7	120	189
650	186	>1100	4.59E-09		900	63	1100	
705	186	46.2	6.25E-07	26.8	13	1	16	30.2
705	124	848.1	3.61E-08	36.8	250	50	275	573
705	110	1475	2.28E-08	19.6	300	?25	400	1075
760	152	4.7	4.72E-06	46.8	1.6	0.4	2	2.7
760	124	28.9	1.71E-06	53	5	0.5	6	22.9
760	103	132.2	5.47E-07	43.7	60	3	70	62.2
760	76	1265	2.78E-08	28.1	450	68	475	790
760	70	4490	7.58E-09	29.5	2600	11.5	4300	190
815	110	14.6	4.44E-06	59.6	2.5	0.2	4	10.6
815	86	37.5	1.03E-06	22.3	12.5	1.8	13.5	24
815	83	72.6		32.1	18	4	20	52.6
815	76	83.2	4.89E-07	23.3	38	4.5	40	43.2
815	62	780.6	1.74E-08	18.1	225	78	250	531
815	52	432	4.67E-08	29	125	2	300	132
815	52	636	3.03E-08	23.6	115	1.7	435	201
815	45	2862	1.11E-08	18.7		14.5	2400	1462
815	41		1.36E-09					
870	76	19.3	1.39E-06	32.2	4.5	2.1	4.5	14.8
870	62	39.3	4.81E-07	15.9	11	5.5	12	27.3
870	48	161	1.5E-07	18.1	44	19	44	117
870	34	451.6	9.92E-08	23.1	270	8	370	81.6
870	31		1.31E-09					
870	26	1728	2.22E-09	16.9		14.9	1005	723
925	48	53	1.75E-07	24	11	14	11	42
925	41	130.5	5.5E-08	23.2	22	38.5	22	109
925	31	292.2	5.56E-08	19.7	62	48	62	230
925	21	>675	1.39E-09		260	675	280	

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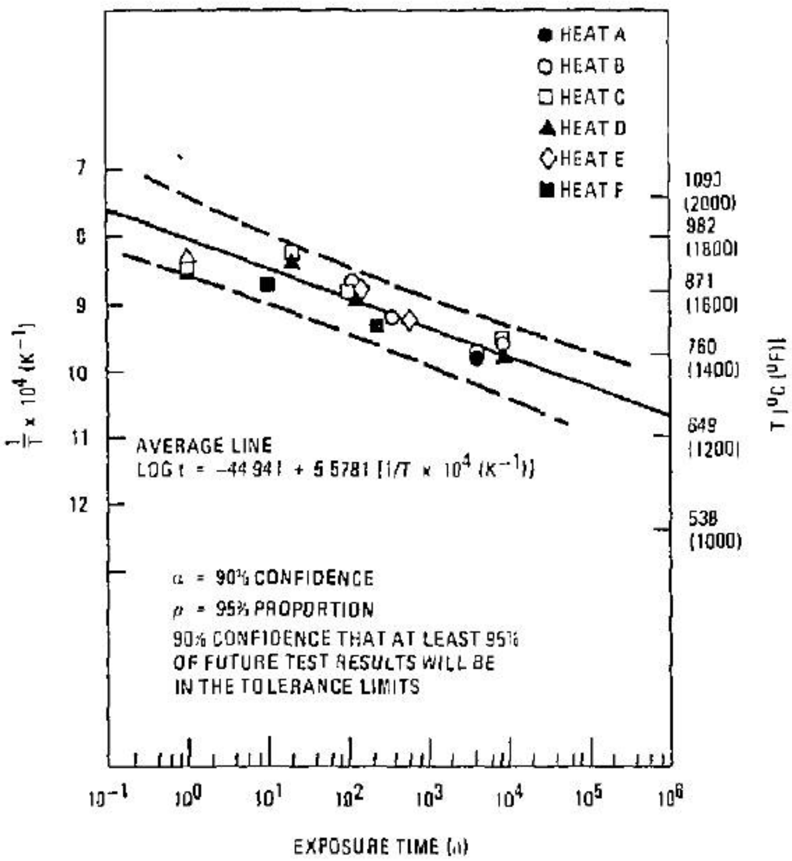

Fig 1: Time temperature envelope for the onset of recrystallisation in 20% prestrained Alloy 800H. (From Roberts, 1981)

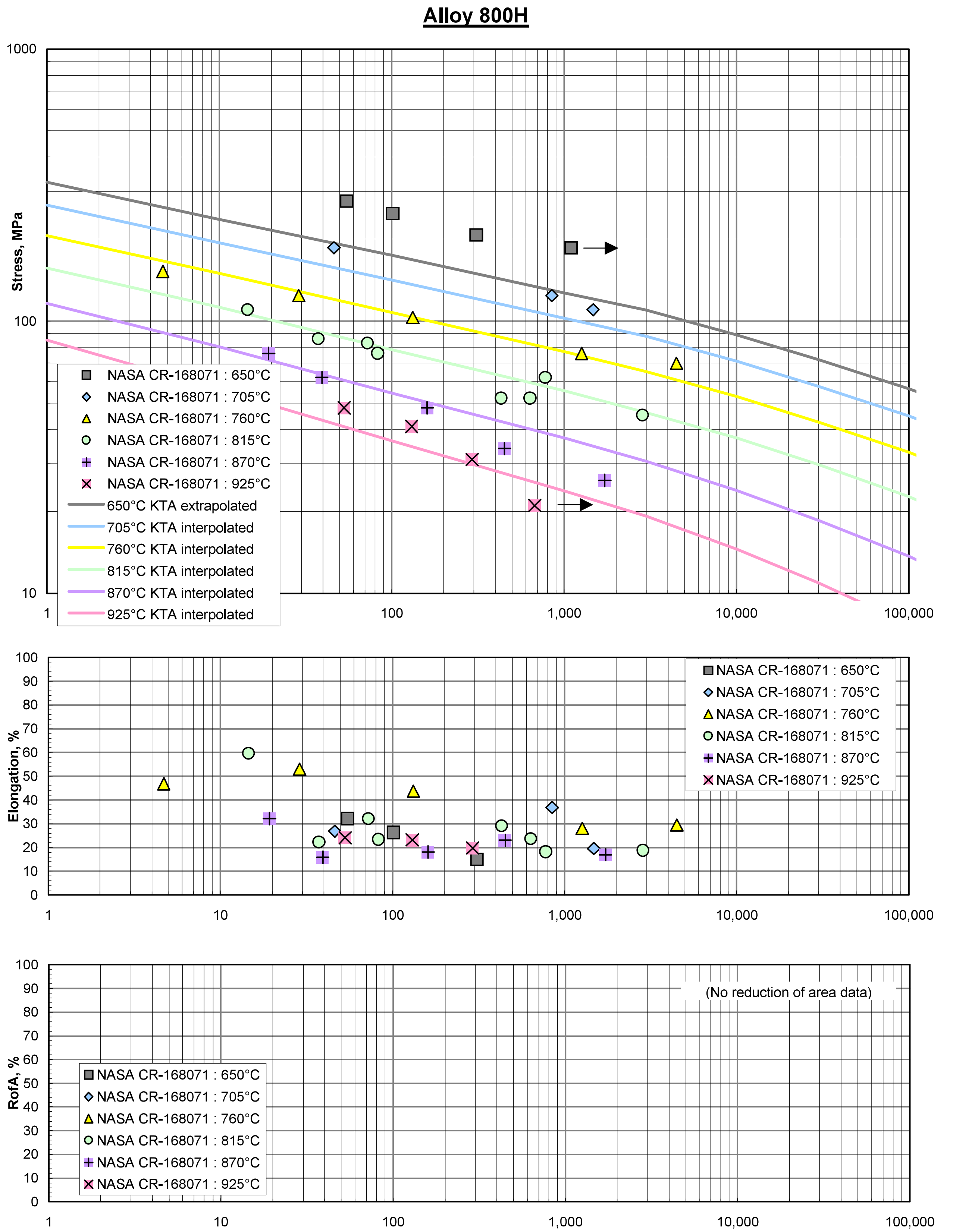

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Fig 2: Stress rupture test data on Alloy 800H thin sheet, air, 650 – 925°C. Arrows indicate unfailed tests. (Extracted from Bhattacharyya, 1982) Interpolated mean strength values from the KTA draft standard are plotted for comparison in the upper part.

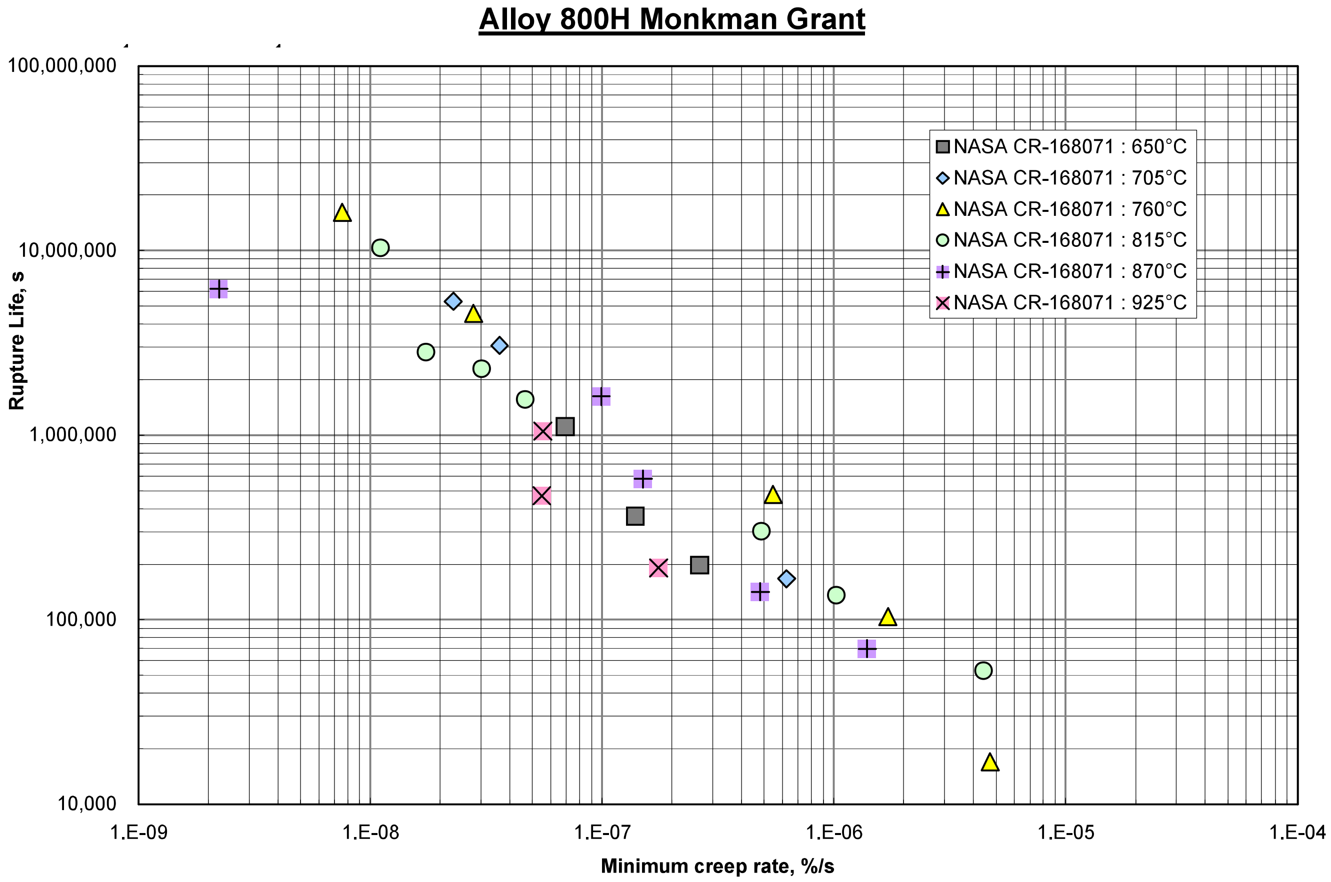

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Fig 3: Monkman-Grant plot of Alloy 800H thin sheet creep rupture test data, air, 650 – 925°C. (Extracted from Bhattacharyya, 1982)

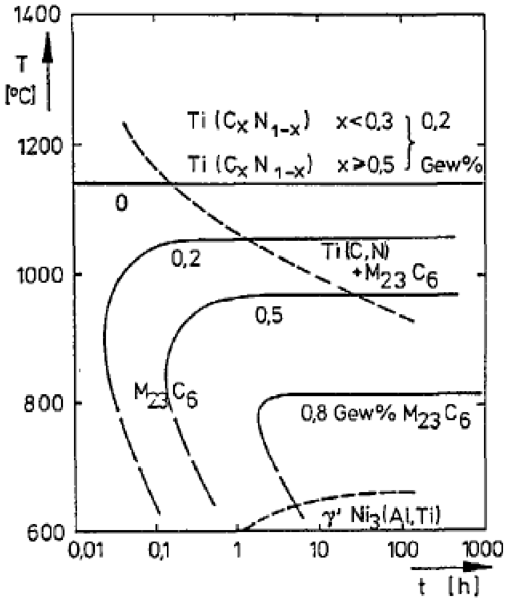

Fig 4: Time-temperature-precipitation diagram for Alloy 800H (From Degischer et al, 1985)

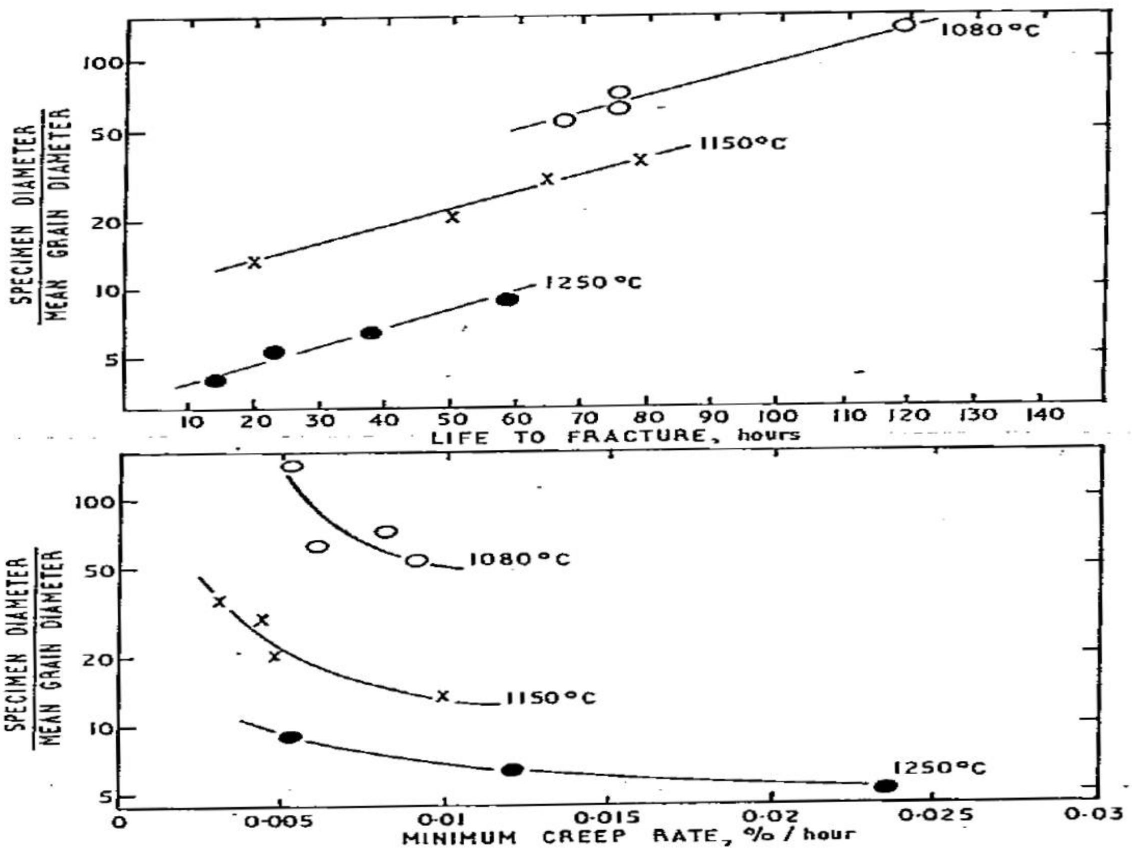

Fig 5: The influence of the ratio of specimen diameter to grain diameter, and solution treatment temperature on the creep properties of wrought nickel alloy (composition similar to Nimonic 90). From Richards, 1968.

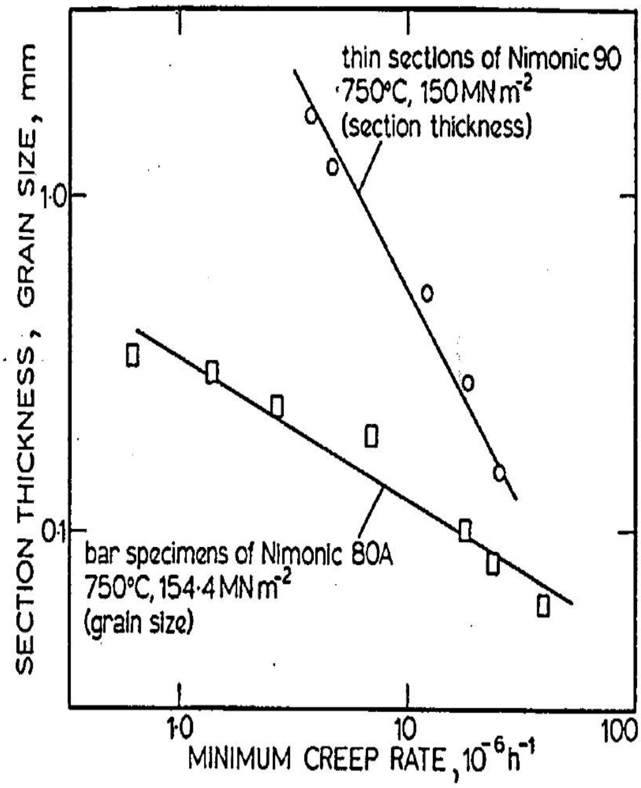

Fig 6: The influence of section thickness or grain size of two wrought nickel alloys (Nimonic 80A, Nimonic 90). From Gibbons, 1981.

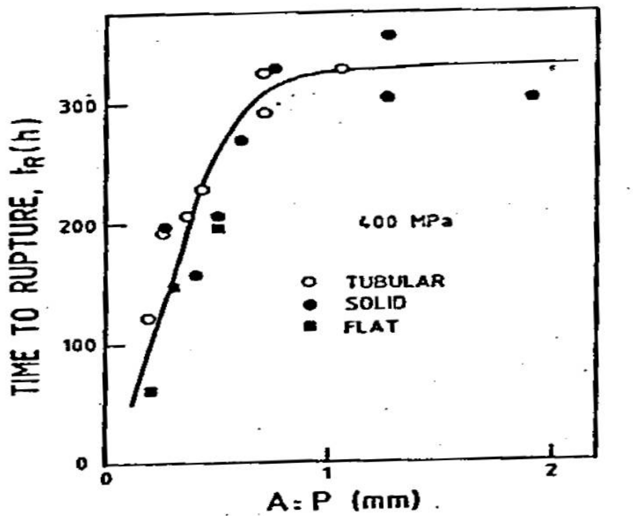

Fig 7: The influence of specimen geometry on Inconel X-750. A is the cross-sectional area and P is the perimeter of tubular, solid and flat samples tested at 700°C. From Pandey, 1989.