Type '300H' Austenitic Stainless Steel Weld Metals for High Temperature Service

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Abstract

There is continuing interest in the use of type '300H' austenitic stainless steel weld metals for a variety of high temperature applications including power generating plant, petrochemical and refinery plant, and other engineering equipment with operating temperatures up to about 800^{O}C. In some cases simple hot tensile strength is the main design requirement, whereas in others, very long term metallurgical stability and creep performance is essential.

A variety of '300H' type weld metals with different composition balances are available, stabilised or unstabilised and with or without molybdenum. Welding process, flux type and residuals may also have a significant influence on elevated temperature properties.

Some new hot tensile and stress-rupture data from a range of candidate weld metals are presented and reviewed in the light of existing published experience. In particular, hot tensile properties, creep-rupture ductility and rupture stress of weld metal deposited by two flux cored wires tested were found to be comparable with published values for bismuth-free consumables.

Recommendations as to the optimum choice of weld metal for specific engineering requirements are suggested.

Keywords: austenitic stainless welds
flux cored 308H/316H
hot tensile tests
stress-rupture tests
bismuth

1. Introduction

Austenitic stainless steels have been and are widely used in the power, petrochemical and nuclear industries for service at elevated temperatures. The combination of high creep rupture strength, high temperature scaling resistance, good general corrosion resistance and reasonable cost is often attractive for critical applications. For example, steels of the 316H type are used for power station steam pipes and superheaters at temperatures in the range 550 to 650^{o}C. In the petrochemical industry, steels of the 304H type are used extensively in refinery catalytic crackers (ref. 1) which operate in the range 500 – 800^{o}C. Such steels are expected to have design lives in excess of 10^5 hours, and in the case of power generation plant, often in excess of 2 x 10^5 hours.

In any of the above installations, welded joints are of major concern, and it is essential that they perform in a similar way to the parent steels without significantly reducing the design life, or compromising the overall integrity of the structure. The weld metals used are generally of similar composition to the parent steels, although consumables of the 308H or 316H type may be used to weld type 347 and 321 stainless steels because of the improved weld metal creep rupture ductility (ref. 2).

The weld metals have a number of features in common, irrespective of the detailed differences in composition of the different types, namely –

a) Carbon controlled in the range 0.04 – 0.08% to ensure adequate creep strength

b) Ferrite controlled, usually in the range 2 – 8FN, to avoid the risk of hot cracking (solidification cracking and/or microfissuring), and at the same time minimise long-term service embrittlement and loss of impact toughness. For the latter reason, the total level of ferrite-promoting elements is normally also restricted.

c) Detrimental minor elements are controlled to enhance long-term, high temperature properties.

A detailed review of the microstructure and composition on the mechanical properties of some AISI 300 series weld metals, including creep properties, has been carried out by Smith and Farrar (ref. 3). This work has shown that high temperature stability and properties are not only strongly dependent upon the carbon, silicon, chromium and molybdenum levels in the weld metals, but are also influenced by residual elements related to consumable design and, in the case of fluxed welding processes, to the type and composition of the flux being used (ref. 4).

Modern fabrication trends, particularly for site construction and repair, are remorselessly moving towards high productivity welding processes. This may be no more than improving the "welder appeal" of the traditional shielded metal arc (SMAW) coated electrode, or it may involve process/consumable changes such as flux cored arc welding (FCAW) to increase both weld metal deposit rates and operator duty cycles.

For those concerned with the development of welding procedures and specification of welding consumables for the manufacture or repair of high integrity, high temperature equipment, any changes from established custom and practice are likely to be challenged with respect to long-term performance. This paper attempts to provide some data and reassurance regarding the high temperature properties of a number of 300 series weldments made using both SMAW and FCAW consumables.

2. Testing Programme

The two main objectives of the testing programme were:

a) To compare the hot tensile properties of a wide range of welding consumables in the temperature range 650°C to 816°C.

b) To establish the creep properties of two flux cored wires over a similar temperature range and compare the results with existing data from consumables for other welding processes and with that from parent steels.

The welding consumables chosen for this programme were all considered representative of current production, and are presented in three groups (Table 1).

The weld metal compositions are given in Table 2, and the groupings are:

2.1 308H types

These are consumables designed primarily for welding 304H stainless steels. They have a controlled carbon content of about 0.05% and simple base composition of 19%Cr and 9%Ni. There are no deliberate molybdenum additions and the Mo values quoted are residual. (Some authorities specify a maximum of 0.25% Mo+Nb+Ti.) All the consumables have deliberate boron additions in the range 20 – 30ppm.

Two coated electrodes are included, one with a rutile coating and one with a basic coating. Compositions are generally similar, except for manganese and silicon levels. One flux cored wire (FCW) is included in this group, with a composition similar to that of the coated electrodes, with the exception of certain residuals arising from differences in raw materials (P, Mo and Cu).

2.2 316H types

These consumables are designed for welding 316H stainless steel, but are also used for 304H, 321H and 347H types. There are six consumables in this group. The leanest are a '16.8.2' sub-group consisting of a solid wire, two coated electrodes and a FCW, all designed with a composition of 16%Cr, 8-9%Ni and Mo in the range 1 to 1.3%. This weld metal composition, originally developed for welding thick section 347 (ref. 5) is specified by a number of oil companies for welding 304H used in refinery catalytic crackers, and aims to exploit the benefits of molybdenum in improving high temperature ductility without significantly increasing the risk of catastrophic oxidation in stagnant air (ref. 1). The compositions are all generally similar, with the exception of residuals arising from raw materials, and boron which is deliberately added to the electrodes and FCW but is a residual 8 ppm in the solid wire. As with 308H types, basic and rutile versions of the covered electrodes were chosen.

The two remaining consumables in the group, both coated electrodes, are a rutile 316H with essentially a standard 316 composition with controlled carbon and residual boron at 10ppm, and a specially designed leaner 316H, low silicon rutile type with an extensive track record of service in high temperature power generating plant (ref 3). This electrode, 17.8.2RCF, has a somewhat higher controlled carbon content and carefully controlled Cr, Ni and Mo levels to give better microstructural stability than standard 316H weld metal after very long periods of exposure at <650^{O}C. In this electrode the boron is increased somewhat to 30-40ppm.

2.3 347H types

Parent materials of types 347H and 321H are still used quite extensively for high temperature plant. However, matching composition consumables are not always the best choice (ref. 2) and for many applications these steels are welded with non-matching 308H or 316H (16.8.2) consumable types. However, for completeness, two coated electrodes have been included in the test programme. Again, they represent rutile and basic types with corresponding differences in manganese and silicon levels. Levels of niobium, for stabilisation, are quite modest at about 8x and 12x the carbon contents respectively, rather than the higher levels of up to 1% allowed by specifications. No deliberate boron additions have been made and the values quoted are residual.

3. Weldments, test specimens and testing procedures

3.1 Weldments

All-weld metal coupons from which the hot tensile and creep test specimens were extracted, were produced generally in accordance with the requirements of AWS specifications. Prior to welding, the 20mm thick low carbon steel plates were buttered with two layers using the consumable to be tested (Table 2) to ensure that only undiluted weld metal was incorporated into the test specimens. Welding conditions were typical of those recommended for the particular consumable and interpass temperature was <150^{O}C. Shielding gas for the two FCW deposits was argon +20%CO_2 and pure argon was used to deposit the TIG weld. In addition to the specimens for high temperature tests, slices were extracted for analysis (Table 2) and for ferrite survey (Table 3).

3.2 Test specimens

The hot tensile specimens were 5mm in gauge diameter, and were machined to the dimensions shown in Fig. 1. The creep specimens were 9mm in gauge diameter and were machined to the dimensions shown in Fig. 2. The diameter of the creep specimens was deliberately made as large as practicable for two reasons: to sample as great a volume of the weld metal as possible and so smooth out the effect of bead-to-bead microstructural variations in the multi-pass weld metal, and to provide a low surface-to-volume ratio which minimises any effects of oxidation.

3.3 Hot tensile testing

The testing procedure was according to EN10002-5 (ref. 6). Testing temperatures were chosen as 650^{O}C (1200^{O}F), 732^{O}C (1350 ^{O}F) and 816^{O}C (1500 ^{O}F) which correspond to temperatures of interest for the power generation industry (up to 650^{O}C) and for the petrochemical industry (up to 800^{O}C). These are conventional benchmark test temperatures in the USA, 816 ^{O}C being the highest temperature for which allowable stresses are specified for 3XXH parent materials in the ASME Code (ref. 7).

3.4 Creep testing

The testing procedure was in accordance with BS 2500 Part 3 (ref. 8). The testing temperatures were the same as for the hot tensile tests, and loads were chosen for each temperature to give expected rupture times of approximately 100h and 1,000h. In addition to rupture time, creep strain was also recorded. However, owing to the specimen size and machine availability, the strains were measured by classical optical extensometry, and the additional processing required to evaluate creep rates precludes reporting in this paper. For the same reason, rupture ductility was reported as reduction of area and not total elongation at fracture.

4. Results and discussions

4.1 Ferrite determinations

Ferrite number (FN) was determined for each of the 11 weld metal slices, both from the composition (Table 2) using the WRC-1992 and DeLong constitution diagrams, and by measurements on each slice at 3 locations, using the Fischer Feritscope. A large number of readings (generally >25) were taken at each location. The results for predicted and measured FN are given in Table 3.

Although most contractors today accept reporting of ferrite as the measured FN value, some continue to specify ferrite according to one of the constitution diagrams, and occasionally both of these requirements may be combined. Lower limits of 2FN or 3FN and upper limits of 8FN or up to 10FN are common for consumables intended for high temperature service. None of the FN values reported here exceeds 6FN, but some fall below 2-3FN according to one or more methods and/or locations. There was no evidence of hot cracking in the low FN deposits. However, additional observations emerged in carrying out the ferrite survey.

Initially, it was found that two of the leanest (16-8-2) weld metals could not be plotted on the current WRC-1992 diagram, so liberty was taken to extend it to include compositions which were evidently not present in the original WRC weld metal database. Also, the DeLong diagram predicted that all the 16-8-2 weld metals were likely to contain martensite, although there was no evidence of this from FN measurements (both martensite and delta ferrite being ferromagnetic). The DeLong martensite boundary was derived from the original Schaeffler diagram, and Kotecki has recently (ref. 9) introduced a corrected ‘quasi-martensite’ boundary to the WRC diagram. This has also been added to the extended WRC diagram shown in Fig. 3, with the locations of the present weld metals indicated – all predicted to be free of martensite.

It was expected that WRC FN would show better correlation with measured FN than DeLong, and this was generally true, particularly regarding the measured values taken at location A on the crown of the last bead, which was linished to remove any oxide/slag film. This corresponds essentially to the situation in the AWS A5.4 reference method for FN determination. The WRC values were usually equal to or slightly below these measured values.

However, a surprising observation was made by careful measurements on the slice faces at location B, in the same bead but orthogonal to A. This location typically indicated 0.5 – 1.5FN higher than the bead crown (A). The same locations (A and B) were next briefly examined using a passive magnetic method (Elcometer Inspector 6F), and the higher values at B were then barely detectable with any certainty.

These observations are not new to the authors, although no relevant published work has been found. Extended discussions are beyond the scope of this paper, but it is provisionally proposed that the Feritscope’s much smaller ‘measurement penetration hemisphere’ is sensitive to the prevailing ferrite orientation morphologies (A versus B) whereas the Elcometer’s larger hemisphere responds with lower resolution to the bulk ferrite volume. Further, systematic deviations in the correlation of these effects with each composition may reflect shifts in ferrite morphology with solidification behaviour.

Setting aside any metallurgical implications, these matters serve to illustrate some of the potential difficulties faced by the manufacturer in designing low ferrite consumables required to meet simple limits when ferrite is examined from different perspectives.

4.2 Hot tensile tests

The results of the hot tensile tests are given in Table 4 and the plots of 0.2% proof strength (PS) versus ultimate tensile strength (UTS) and % elongation versus % reduction in area (R of A) are given in Figures 4a and 4b respectively. The ductility diagram shows a high degree of scatter particularly at R of A values in excess of 50% and all the 347H types show significantly lower values than the 308H and 316H types. The strength diagram shows good correlation between 0.2% PS and UTS particularly for the 308H and 316H types. The 347H results all tend to lie on the upper edge of the scatter band.

In order to present the data in a more visual form, all the hot tensile and ductility results are shown as bar charts in Figure 5a and 5b respectively. Within a given alloy group there is a high degree of consistency of tensile strength at all three test temperatures. The only real exceptions are the higher alloyed 316H types, namely E316H and 17.8.2.RCF, which have higher molybdenum and chromium contents, and, in the case of the 17.8.2.RCF, a higher carbon content than the remaining "lean" 16.8.2 types. This increase in strength is present at 650^{O}C and 816^{O}C, but less apparent at 732^{O}C. The increase is significant at 650^{O}C, a temperature of interest to the power generation industry and at this temperature these 316H types are at least as strong as the 347H types. However, this performance does not hold up at the higher temperatures, and the 347H types demonstrate superior strength at 732^{O}C and 816^{O}C.

There appears to be no significant effect of consumable type – solid wire, coated electrode or FCW – on strength, nor in the case of coated electrodes, any effect of coating type, whether it be basic, rutile or acid rutile in character.

Similar comments can be made regarding ductility, with all the 308H and 316H showing consistent behaviour at all three test temperatures. No great significance should be attached to minor differences in % R of A, since values in the 50 – 80% range are quite difficult to measure, and the possible error of any individual reading is about ±5%. However, it can be clearly stated that all the 308H and 316H types, irrespective of consumable form or coating type, gave ductility values in excess of 50% R of A at all three test temperatures.

On the other hand, the E347H types exhibited quite different behaviour, with significantly lower ductilities at all test temperatures, and serious further reductions at the higher testing temperatures, down to about 20% R of A at 816^{O}C. The data for the 347H and 17.8.2 weld metals are in good agreement with previously published work by Boniszewski (ref. 2) which emphasised the benefits of using molybdenum-bearing weld metals to improve hot ductility.

4.3 Creep rupture tests

The results of the creep rupture tests are given in Table 5. The first two tests carried out at 650^{O}C were unwittingly loaded to levels rather close to the 0.2% proof stresses of the respective materials at that temperature and consequently rupture occurred after very short periods. The tests are closer to hot tensile tests than time-to-rupture tests and therefore failure was premature and the results are of limited value. The remainder of the stress levels used and the resulting rupture times are all valid. Creep rate behaviour is not discussed in this paper, but a representative creep curve for SC308H at 816 ^{O}C and 44MPa is shown in Figure 6.

The stress-rupture results are shown as Larson-Miller plots to enable comparisons to be made with relevant parent materials and weld metals. Figure 7 shows the data obtained from both flux cored wires, SC308H and SC16.8.2, plotted on the same diagram as a single cast of 304 with 0.04%C. This material is considered typical of the parent steel for which these consumables would be used for fabrication in high temperature refinery plant. It can be seen that the stress-rupture performance of the two weld metals is similar, and plot together as a single linear curve with slope very similar to the 304 but at higher stress. These trends suggest that there are no radical shifts or differences in the characteristics of microstructural evolution of parent and weld metals for the tested durations and temperatures. The deviation of weld metals at >200MPa is considered to be the result of non-valid test conditions already explained. The lower rupture stress for the 304 may be partly owing to a lower C+N level (0.07%) compared with the weld metals (~0.1%).

In Figure 8, the same data for SC308H and SC16.8.2 is plotted for comparison with the minimum 10^5h rupture stress curves for relevant 300H type parent materials. These curves represent the baseline from which ASME maximum allowable stresses are derived (ref. 7), for service at temperatures up to 816^{O}C. Again, the weld metals generally match or overmatch the minimum rupture stresses of all the parent materials, particularly at the higher Larson-Miller parameter values. This may indicate that in cross-weld tests, failure in the weld metal would be more likely in short-term tests than in longer term tests.

With regard to rupture ductility as measured by %R of A, the molybdenum-bearing SC16.8.2 is generally superior to the SC308H. This is most apparent at the higher temperatures of 732^{O}C and 816^{O}C, and after longer rupture times.

Additional comparisons were also made with in-house and published data to examine the stress rupture characteristics of relevant weld metals of 308H and 316H/16.8.2 type composition. On the basis of Larson-Miller evaluation, the present FCW data were found to fall within the range expected from other processes, notably SMAW. TIG weld metal of type ER347H was generally stronger and ER316H slightly weaker, while ER308H was weakest in terms of rupture stress for a given L-M parameter. The latter might be owing partly to compositional balance and/or the absence of some residuals which contribute to optimum high temperature performance.

4.4 Comparison with other published results

Modern small diameter stainless flux cored wires were developed and have been most extensively used in Japan. It is, therefore, not surprising that most of the literature covering high temperature properties emanates from this source. Failures of welded joints made with 308H type FCW's have been reported and, although it is recognised that the long term performance of such joints may well be influenced by delta ferrite transformation to sigma phase and/or carbide precipitation, it has been stated that the most significant effect was caused by bismuth (refs. 10, 11), a very small quantity of which is often added (as bismuth oxide) to the flux core to improve slag removal and weld bead appearance. Although this is harmless in normal applications, its use in this case is rather surprising in view of the long-known influence of bismuth on hot ductility (ref. 12), and its adverse effect on FCW weld metals for high temperature service has now been investigated by a number of workers (refs. 13, 14, 15). The main conclusions were as follows:

- Intergranular cracks can occur after short service periods at temperatures in excess of 700 ^{O}C.
- Hot ductility at temperatures above 650 ^{O}C is greatly reduced.
- Fracture surfaces show clear evidence of liquated films and Auger analysis confirmed the presence of bismuth on the surfaces.
- Crack sensitivity and drastic loss in ductility is caused by melting of bismuth oxide, leading to grain boundary liquation and hence a reduction in grain boundary strength at temperatures in the range 650 - 800 ^{O}C .
- Creep crack growth rate is greatly increased by the presence of bismuth and ductility was correspondingly reduced.

The FCW's described in this paper are specifically intended to be used for high temperature applications and are therefore designed to provide satisfactory de-slagging without the use of bismuth oxide. One petrochemical industry authority has already proposed a limit of 0.002%Bi for FCW intended for service at >675^{O}C (ref. 16).

The detrimental effect of bismuth on hot tensile ductility is clearly shown in Fig. 9(a) (from ref. 14) and the 308H data from the present work have been superimposed and can be seen to be in good agreement with the more ductile Bi-free data. Fig. 9(b) from the same source shows that hot tensile strength is unaffected by the presence of bismuth and, again, results from the present work follow the same trend as the other data.

Similar comparisons can be made with the stress-rupture data. Figure 10 from Toyoda et al (ref. 13) shows the results of an extensive testing programme on a wide range of stainless FCW's covering a range of bismuth levels plus other variations in oxygen, nitrogen and boron. Of particular relevance to this work are the points plotted for weld metal C, which is from a 308H FCW with low bismuth (<0.001%). The stress versus time to rupture results from the present work, for both 308H and 16.8.2 types, are shown plotted on Figure 10(a) and, allowing for the difference in test temperature (732 ^{O}C against 750 ^{O}C) it can be seen that there is very little variation between the 308H and 16.8.2 types. Figure 10(b) shows %R of A against time to rupture and, again, the results for the 308H types are very similar except for the weld with bismuth. However, the SC 16.8.2 weld metal shows slight increase in ductility with time to rupture at 732/750 ^{O}C, which supports the view that molybdenum is beneficial in providing improved rupture ductility. In fact, at all three temperatures in the current work, the SC 16.8.2 weld metal exhibits rupture ductility equal to or better than the SC 308H type, as noted earlier.

5. Conclusions

The hot tensile properties, at 650 ^{O}C, 732^{O}C and 816 ^{O}C, of a number of controlled carbon, low ferrite austenitic 300 series weld metals, have been evaluated. In addition, creep testing at the same temperatures was carried out on welds made from two bismuth-free flux cored wires. The conclusions, which are summarised below, are restricted to those drawn directly from the experimental results and comparisons with other published work. No attempt has been made at this stage to investigate subtle differences in composition, or the effects of temperature on microstructure. It is hoped that the investigation will continue using specimens now available and that this work will form the subject of a future paper, to include creep behaviour in addition to the stress-rupture properties reported in the present paper.

i) Reasonably good agreement was found between ferrite directly measured on the weldments as FN and values predicted from composition using the WRC diagram and the DeLong diagram. However, it was necessary to extend the WRC diagram to include the leaner 16.8.2 weld metals. The correlation was better with the “extended” WRC diagram than with DeLong.

ii) All the 308H and 316H type weld metals showed similar hot tensile properties at the temperatures tested. Small differences in strength were attributed to differences in composition and all the weld metals gave %RofA values in excess of 50%. There was no evidence of any influence of consumable form, i.e. solid wire, coated electrode or flux cored wire, nor – in the case of the coated electrodes – in coating type, basic or rutile.

iii) The two 347H type weld metals exhibited higher strengths and lower ductilities than the 308H/316H types, particularly at the higher temperatures of 732 ^{O}C and 816 ^{O}C.

iv) The welds made with the two flux cored wires exhibited stress-rupture properties which generally exceeded parent material values and were similar to those from previous work using coated manual metal arc electrodes, which have a track record of successful useage in high temperature process plant.

v) The hot tensile ductility and stress-rupture ductility properties of the two flux cored wires were comparable with published values from Japan obtained from <u>bismuth free</u> consumables.

vi) On the basis of the data available, it is concluded that 308H/316H (16.8.2) type flux cored wires can be used with confidence in the fabrication and repair of high temperature plant as high productivity alternatives to more conventional welding processes, provided that bismuth is <u>not</u> included as part of the flux design. A maximum level of 0.002% Bi has been proposed by one authority.

6. Acknowledgements

Thanks are due to Julian Sanderson and Zhuyao Zhang, who were responsible for the design of the welding consumables evaluated in the testing programme, to Mark Golding for preparing the extended WRC diagram, and to colleagues in the Welding Department who produced the necessary test plates.

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Table 1

List of welding consumables tested

GENERIC TYPE	BRAND NAME	CODE	AWS	SIZE mm	FLUX TYPE
'308 H'	Ultramet 308H	UM 308H	E308H-16	4	Rutile
	Ultramet B308H	UM B308H	E308H-15	4	Basic carbonate-fluoride
	Supercore 308H	SC 308H	E308HT0-4	1.2	Rutile FCW
'316 H'	ER16.8.2	ER16.8.2	ER 16-8-2	2.4	Solid TIG wire
	Supermet 16.8.2	SM 16.8.2	E16-8-2-17	4	Rutile-aluminosilicate
	E16.8.2.15	E16.8.2.15	E16-8-2-15	4	Basic carbonate-fluoride
	Supercore 16.8.2	SC 16.8.2	N/A	1.6	Rutile FCW
	17.8.2.RCF	17.8.2RCF	(E16-8-2-16)	4	Rutile (low silica)
	E316H-16	E316H-16	E316H-16	3.2	Rutile
'347 H'	Ultramet 347H	UM 347H	E347-16	4	Rutile
	Ultramet B347H	UM B347H	E347-15	4	Basic carbonate-fluoride

Table 2

Weld metal compositions, % (boron in parts per million)

Consumable	C	Mn	Si	S	P	Cr	Ni	Mo	Cu	Nb	Ti	V	N	B ppm
UM 308H	0.045	1.08	0.70	0.015	0.011	19.0	9.4	0.04	0.04	0.03	0.012	0.06	0.088	29
UM B308H	0.049	1.41	0.26	0.010	0.010	18.4	9.1	0.08	0.06	0.02	0.019	0.06	0.060	20
SC 308H	0.054	1.53	0.55	0.010	0.029	18.1	9.2	0.14	0.17	0.01	0.040	0.06	0.052	26
ER16.8.2	0.054	1.43	0.44	0.006	0.009	15.4	8.6	1.15	0.02	0.01	0.005	0.02	0.049	8
SM 16.8.2	0.055	1.09	0.57	0.019	0.015	15.8	8.0	1.07	0.06	0.01	0.011	0.04	0.063	23
E16.8.2.15	0.047	1.80	0.36	0.008	0.015	16.0	8.5	1.07	0.08	0.03	0.020	0.04	0.031	25
SC 16.8.2	0.045	1.17	0.53	0.011	0.028	16.1	8.9	1.13	0.16	0.01	0.037	0.05	0.050	28
17.8.2.RCF	0.066	2.10	0.28	0.009	0.020	17.5	8.8	2.02	0.08	0.02	0.019	0.06	0.093	38
E316H-16	0.051	1.37	0.87	0.016	0.018	17.9	11.3	2.11	0.07	0.02	0.015	0.05	0.091	10
UM 347H	0.048	0.94	0.74	0.016	0.018	19.2	9.7	0.09	0.07	0.42	0.014	0.05	0.085	7
UM B347H	0.040	1.51	0.31	0.009	0.011	18.4	9.3	0.03	0.04	0.53	0.023	0.05	0.061	6

Analysis was determined on the undiluted mid-section of slices extracted from the test coupons, using a Spectrolab M8 optical emission spectrometer.

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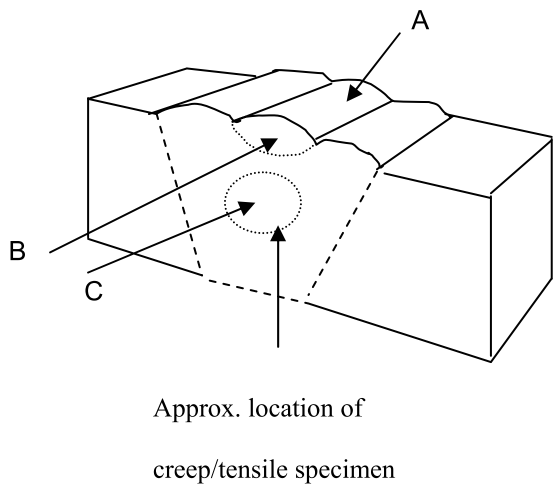
Table 3

Weld metal ferrite determinations

Consumable	Predicted FN		Measured FN (see sketch)		
	WRC	DeLong	A	B	C
UM 308H	3.0	4.3	3.2	3.4	4.0
UM B308H	3.1	2.2	2.8	4.5	2.8
SC 308H	2.1	2.7	4.1	4.8	3.8
ER 16.8.2	0.2 (1)	-1.2 (2, 3)	1.2	1.5	0.4
SM 16.8.2	1.1 (1)	0.8 (3)	1.4	2.1	1.5
E16.8.2.15	2.3	1.2 (3)	2.3	3.6	2.7
SC 16.8.2	1.2	1.2 (3)	1.9	3.1	1.9
17.8.2.RCF	3.6	2.1	4.8	6.2	3.9
E 316H-16	1.4	2.8	1.4	3.6	2.2
UM 347H	3.5	5.8	4.0	4.5	4.5
UM B347H	4.0	3.2	4.3	3.7	5.2

Location of measurement:
A = Top surface of last bead
B = Section of last bead
C = Mid section of weldment



(1) Estimated by extension of WRC-92 diagram to below 17Cr_{eq} (see fig. 3)

(2) Estimated ‘negative FN’ by extrapolation

(3) DeLong (incorrectly) predicts that all the ’16.8.2’ types lie below the austenite + martensite boundary line.

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Table 4. Results of hot tensile tests

Code	0.2% PS MPa			UTS MPa			Elong 5d%			R of A %		
Temperature ^{O}C	**650**	**732**	**816**	**650**	**732**	**816**	**650**	**732**	**816**	**650**	**732**	**816**
UM 308H	234	187	156	297	231	181	28	51	53	55	63	64
UM B308H	223	168	111	298	225	154	24	48	47	60	63	54
SC 308H	213	177	140	287	222	163	30	46	40	58	69	74
ER16.8.2	221	178	147	315	241	173	31	36	42	67	69	65
SM16.8.2	225	179	126	310	232	161	28	47	43	52	59	55
E16.8.2.15	216	187	132	294	230	165	27	36	57	61	70	75
SC16.8.2	207	180	134	290	224	160	30	44	39	66	68	79
17.8.2.RCF	287	197	147	369	274	191	28	44	53	55	61	75
E316H-16	264	204	152	352	268	197	32	43	54	58	53	60
UM 347H	283	269	206	354	308	233	19	20	7	47	38	23
UM B347H	263	265	223	354	311	248	18	14	5	43	30	19

Room Temperature mechanical properties – typical values for the above weld metals are within the ranges shown to the right:

The 347H types tend to be at the upper end of the strength range.

0.2%PS	420 – 520 MPa
Tensile Strength	620 – 680 MPa
Elongation 5d	35 – 45%
R of A	40 – 50%

Table 5. Results of creep tests (stress-to-rupture)

SC 308H				
Series	Temp ^{O}C	Stress MPa	Time, h	R of A %
'100h'	650	210	9	68
	732	115	124	63
	816	55	227	37
	871	35	198	3.3
'1000h'	650	160	1037	47
	732	85	750	23
	816	45	749	14

SC 16.8.2				
Series	Temp ^{O}C	Stress MPa	Time, h	R of A %
'100h'	650	220	10	68
	732	130	39	64
	816	65	78	38
	871	35	303	4.5
'1000h'	650	160	1027	58
	732	85	488	71
	816	45	861	23

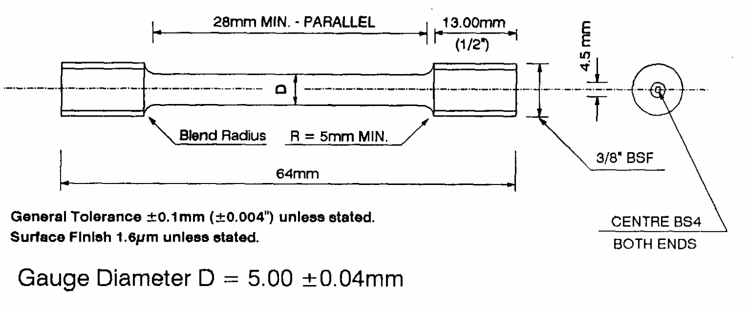


Fig. 1: Dimensions of hot tensile test specimens

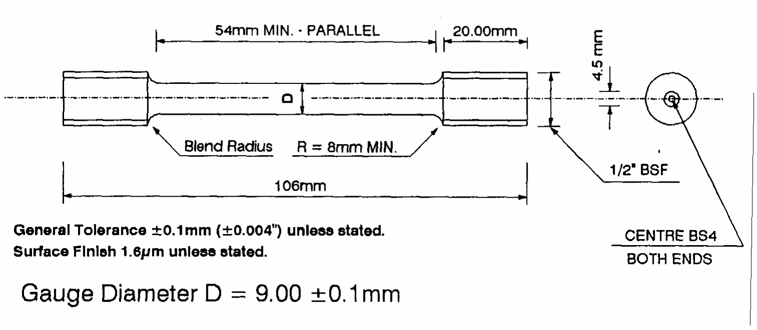
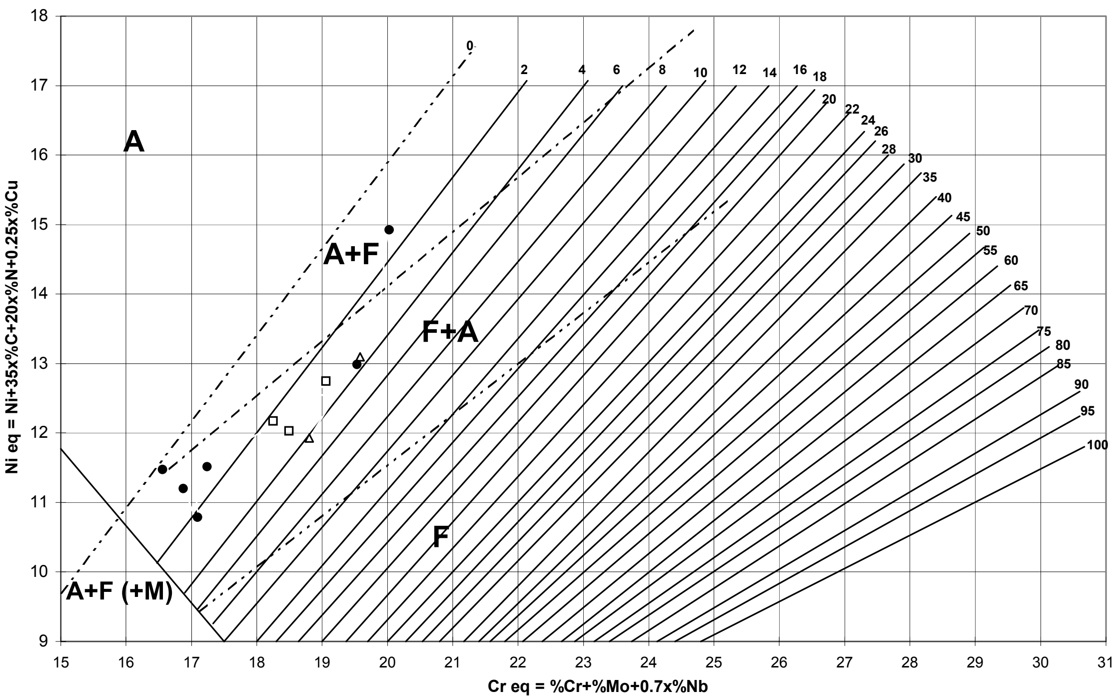
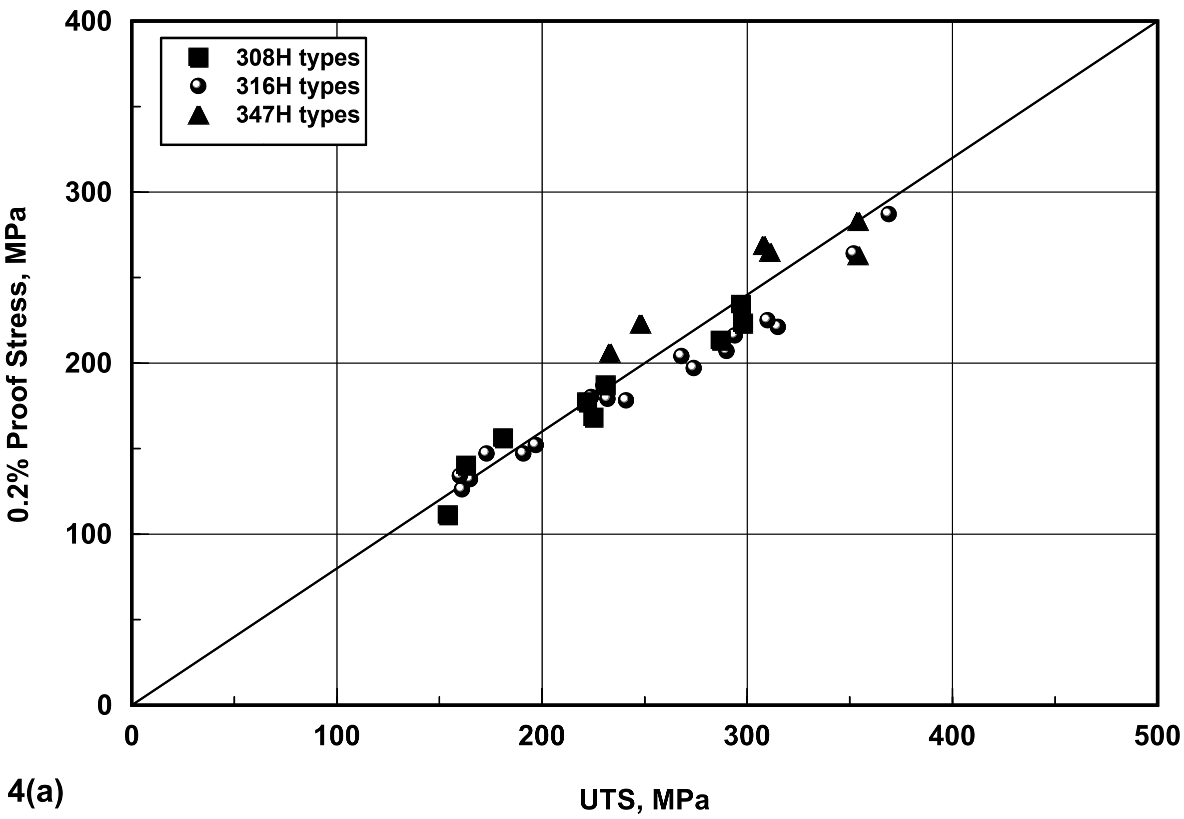


Fig. 2: Dimensions of creep test specimens

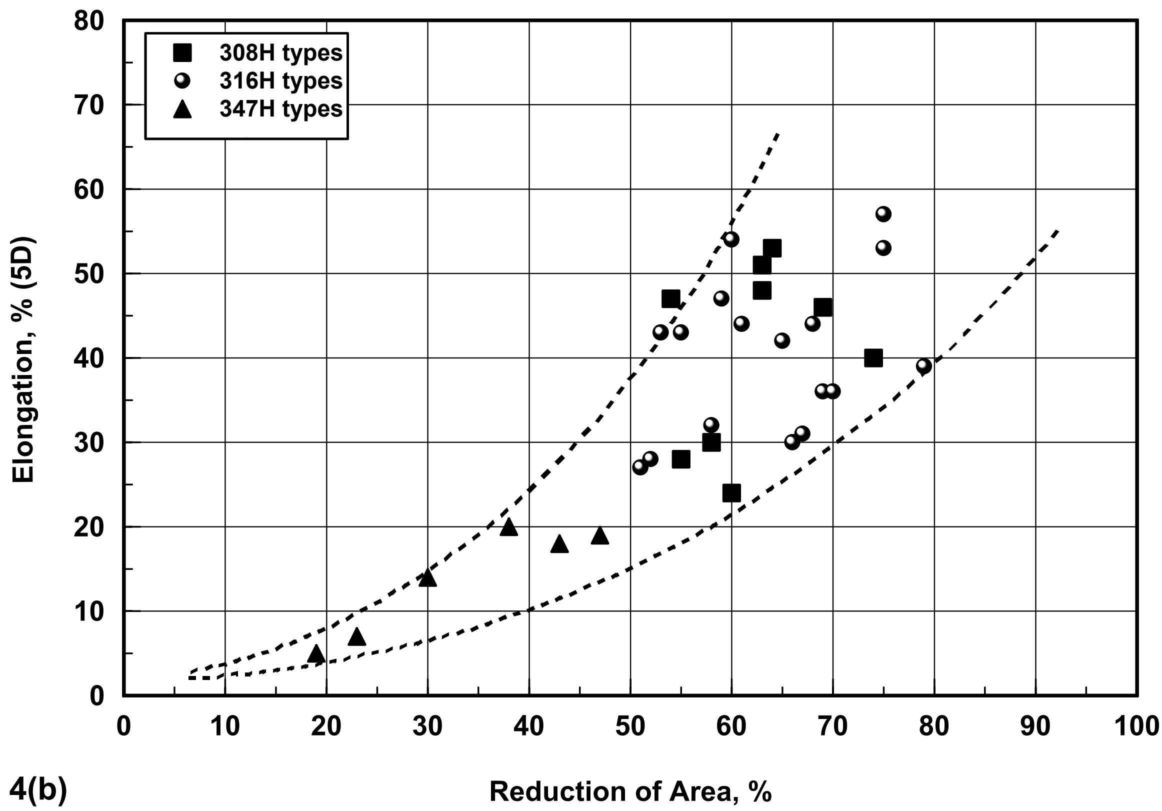


Symbols: □ = 308H, ● = 316H, △ = 347H

Fig. 3: WRC-1992 diagram modified to extend Creq down to 15 and with the addition of Kotecki's quasi-martensite boundary. The location of all the weld metals is shown, including 16-8-2 'lean 316H' types.



4(a)



4(b)

Fig. 4: Hot tensile property relationships of weld metals
(a) 0.2% Proof stress versus tensile strength
(b) Elongation versus reduction of area

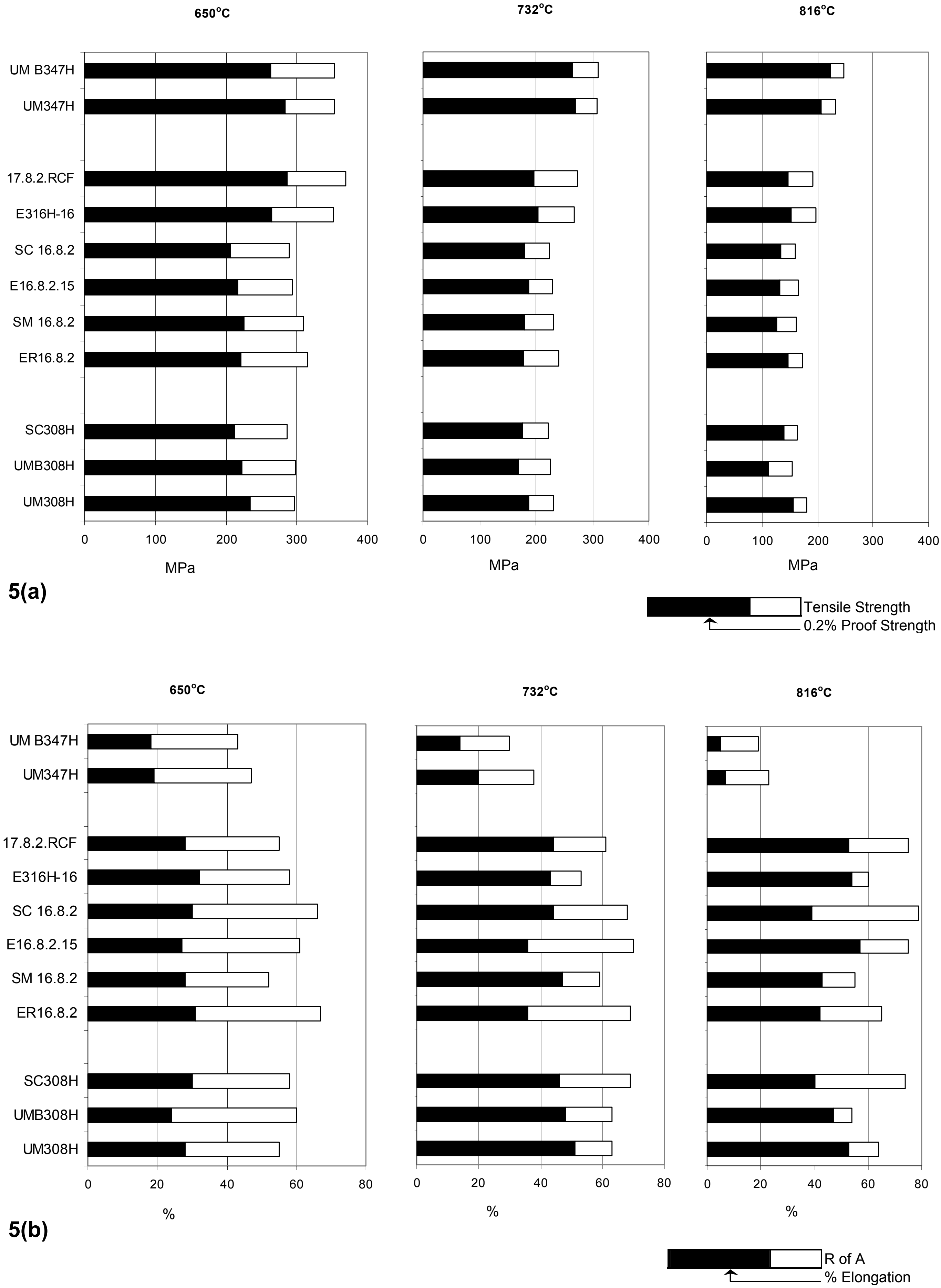


Fig. 5: Hot tensile properties of the weld metals tested
(a) Tensile strength
(b) Ductility

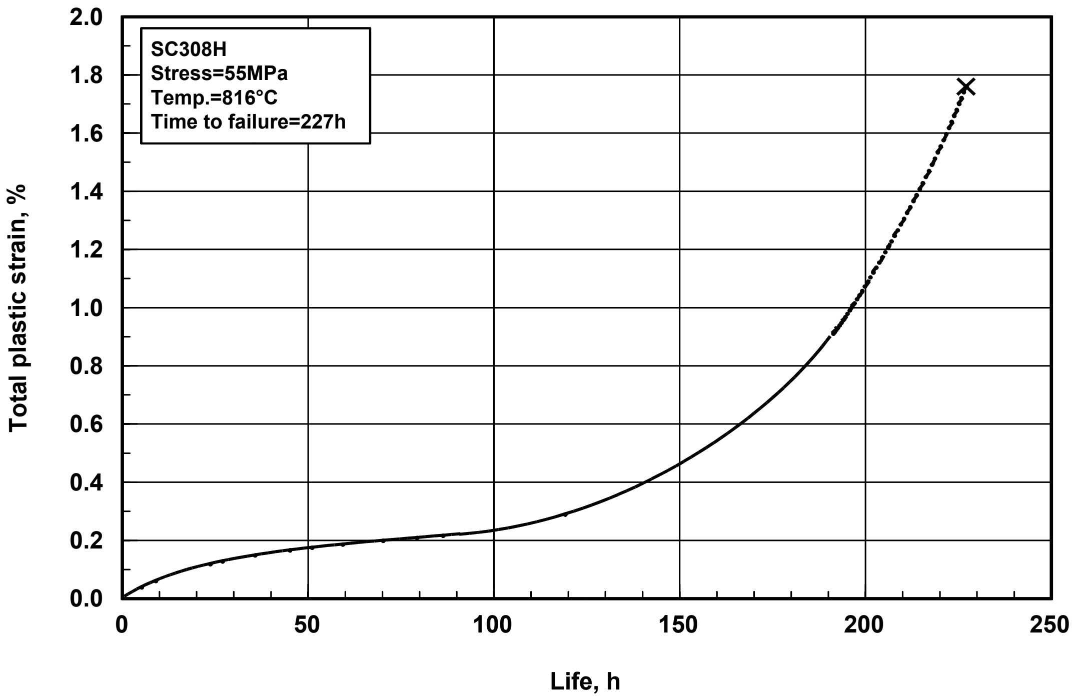


Fig. 6: Representative creep curve for SC308H at 816°C

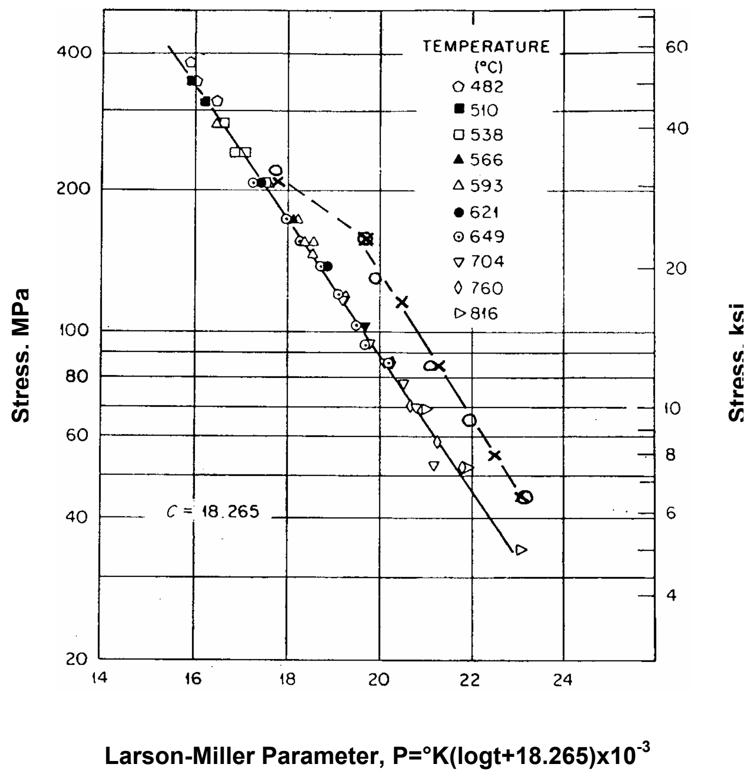


Fig. 7: Larson-Miller plot of stress-rupture data obtained for SC308H and SC16.8.2 FCW weld metal compared with data for a single cast of 304 with 0.04%C (ORNL data from Swindeman, ASME MPC-1, 1975, pp1-29). Additional symbols: X=SC308H, O=SC16.8.2.

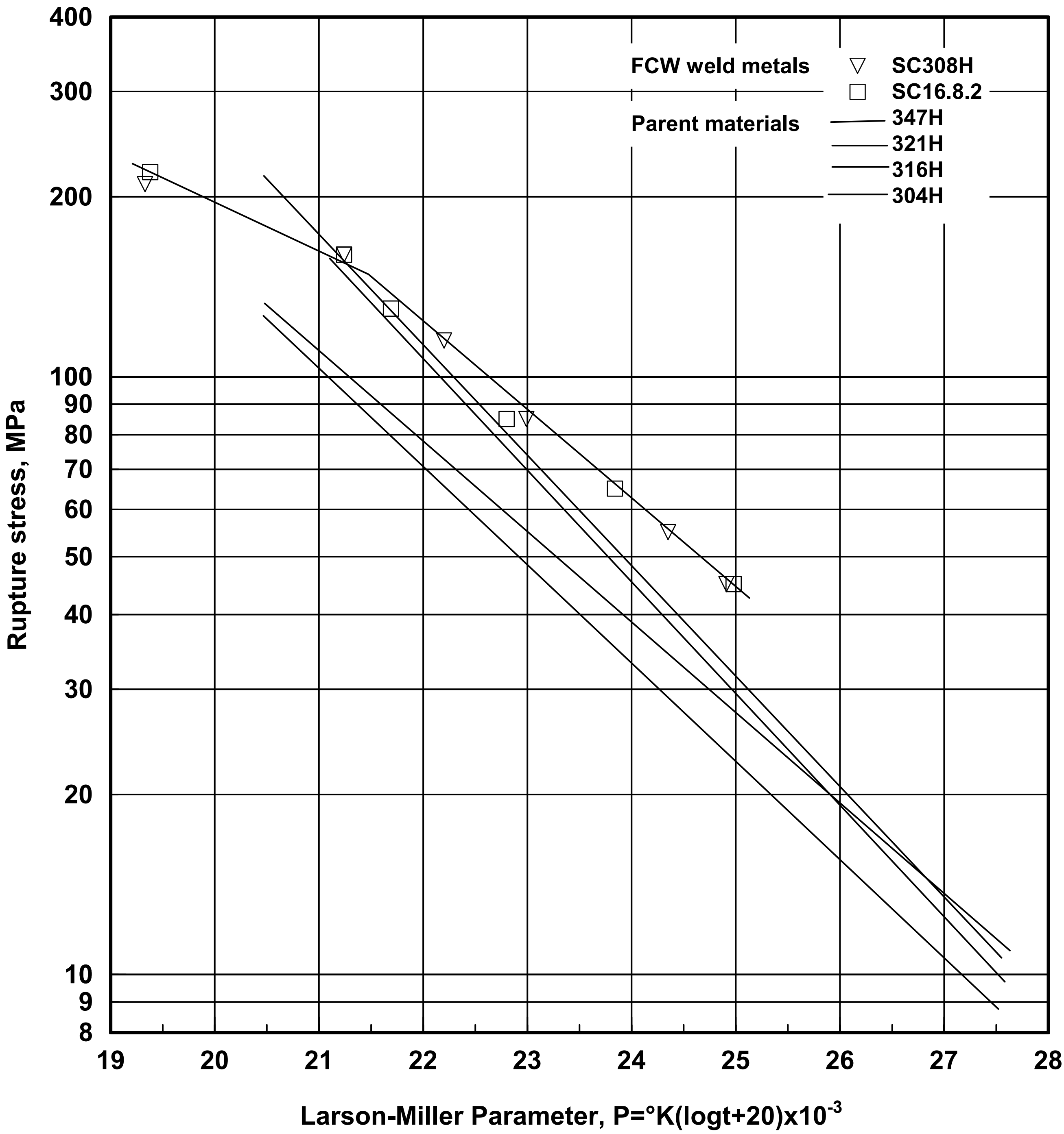
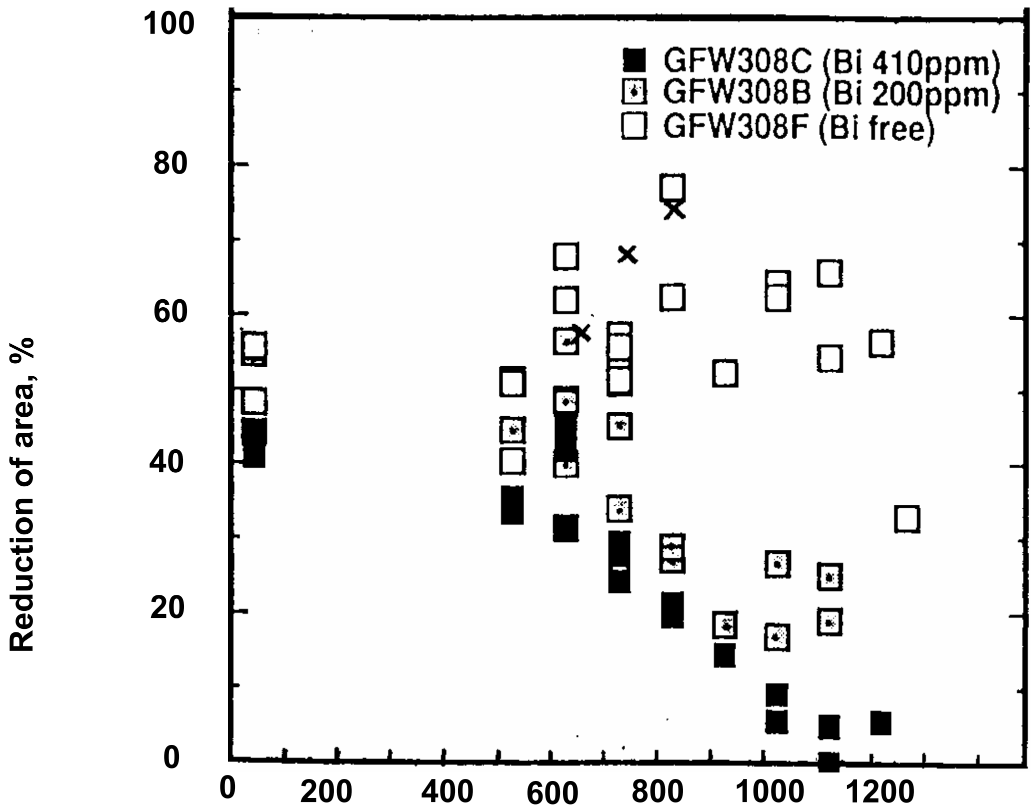
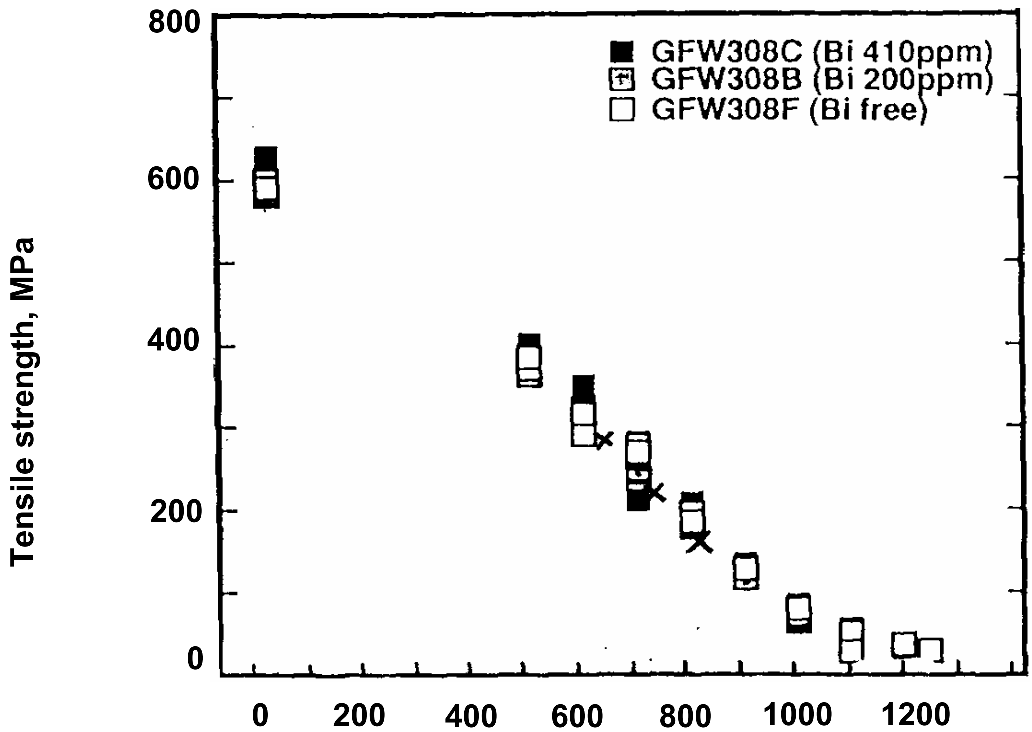


Fig. 8: Larson Miller plot of stress-rupture data obtained for SC308H and SC16.8.2 FCW weld metal compared with minimum 10^5h rupture stress for relevant austenitic 3xxH parent materials (from ASTM DS 5S2)

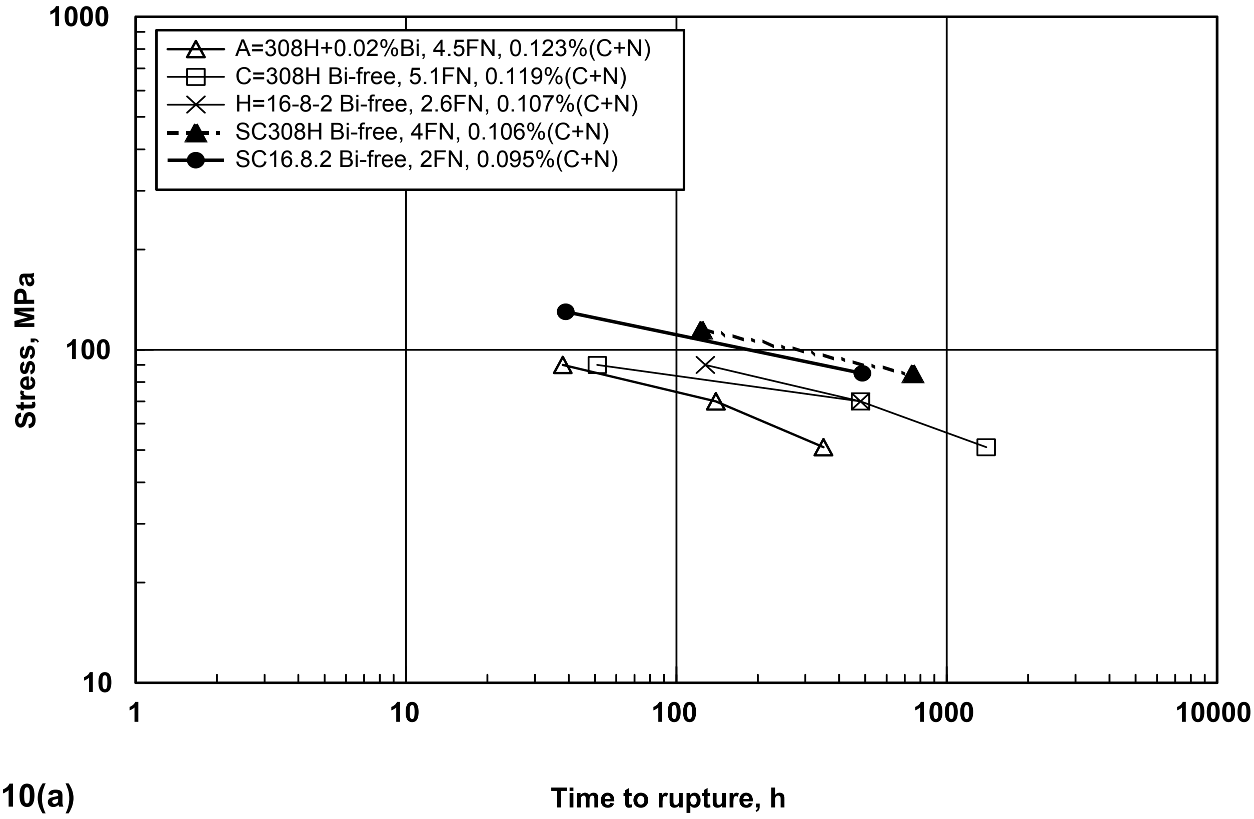


9(a) Temperature, °C

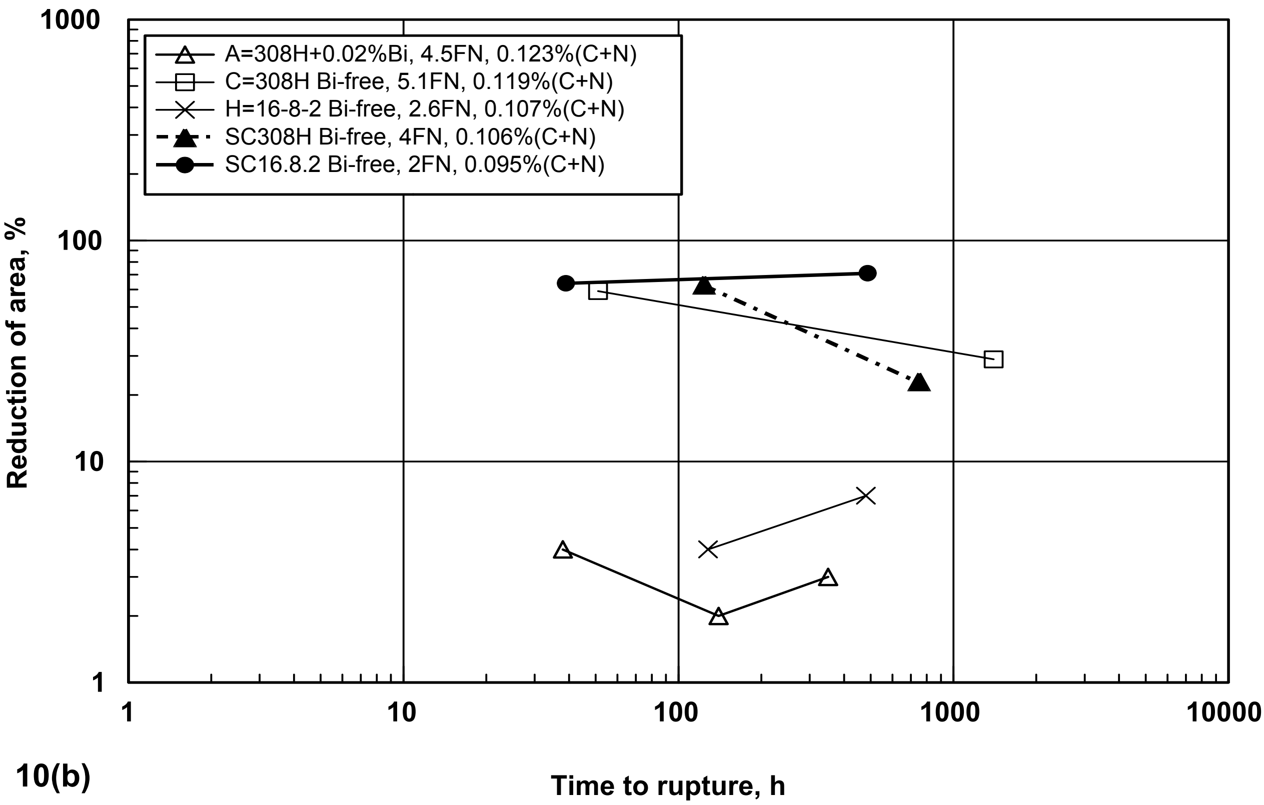


9(b) Temperature, °C

Fig. 9: Hot tensile properties of 308H FCW weld metals with varied Bi level, from Nishimoto et al (ref 14). Data for SC308H in the present work (symbol=X) are added to show similar Bi-free behaviour, (a) ductility, (b) tensile strength



10(a)



10(b)

Fig. 10: Comparison between stress-rupture data at 750°C from Toyoda et al (ref 13) for some FCW weld metals with and without Bi additions, and data obtained at 732°C for SC308H and SC16.8.2 in the present work, (a) rupture stress , (b) rupture ductility