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REPORT

**THE EFFECTS OF AGING
ON DISSIMILAR METAL WELDED JOINTS
WITH MULTIPASS WELDING OF HASTELLOY X
AND HASTELLOY W FILLER METAL**

D. R. Ireland

July, 1966

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by

D. R. Ireland
Engineering Materials and Mechanics Section
Engineering Development Department

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INTRODUCTION

A program was started to evaluate the age embrittlement of Hastelloy X and Hastelloy W weld metals in multipass-welded joints of Hastelloy X-280 to Haynes 25 and 316 SS. This program is evaluating the effects of aging for 500, 1000, and 5000 hr at 1000, 1200, 1400, and 1600 °F on the mechanical properties of Hastelloy X weld metal, Hastelloy W weld metal, Hastelloy X-280 base metal and Haynes 25 base metal. This report describes the results of bend tests, tensile tests, hardness tests, and metallographic examinations of unaged material and material aged 500 and 1000 hr. The primary purpose of this investigation is to determine which of the two weld metals, Hastelloy X or Hastelloy W, are best suited for prolonged exposure to temperatures of 1000 to 1600 °F. The ductilities of the weld metals in multipass-welded joints of Hastelloy X-280 to 316 SS and Hastelloy X-280 to Haynes 25 were chosen for this investigation. The 316 SS is a well documented material and its properties are not evaluated in this study.

Both of these weld metals have been successfully used for joining dissimilar nickel-base alloys.⁽¹⁾ Hastelloy W was specifically developed as a welding filler wire for use in joining dissimilar nickel-base alloys, nickel-base alloys to cobalt-base alloys, and nickel-base alloys to stainless steels. The two weld metals and the base metals evaluated in this program have previously been shown to suffer from age embrittlement when exposed to temperatures of 1200 to 1600 °F.⁽²⁻⁹⁾

SUMMARY AND CONCLUSIONS

Within the limits of this investigation and on the basis of the information obtained to date, Hastelloy W weld metal appears to be better suited to prolonged exposure to temperatures of 1000 to 1600 °F than Hastelloy X weld metal.

This conclusion is based on the test results of material aged 500 and 1000 hr at 1000, 1200, 1400, and 1600 °F.

As expected, both weld metals age embrittle (lose ductility) with exposure to temperatures of 1000 to 1600 °F. This age embrittlement of the weld metals goes through a cycle where the aging has its maximum effect on ductility and then overaging results in recovery of ductility.

Both weld metals reach similar minimum room temperature ductilities of approximately 2%. The aging is more sluggish in Hastelloy W than in Hastelloy X. Hastelloy W reaches its minimum room temperature ductility when aged 500 hr at 1400 °F, while Hastelloy X reaches it when aged 500 hr at 1200 °F.

The minimum elevated temperature ductilities of the two weld metals are quite different. Hastelloy X has a minimum ductility of 5% at 1200 °F, when aged 500 hr at 1200 °F; Hastelloy W has a minimum ductility of 10% at 1400 °F, when aged 500 hr at 1400 °F.

The base metals, Hastelloy X-280 and Haynes 25, also age embrittle with exposure to temperatures of 1000 to 1600 °F. Hastelloy X-280 reaches its minimum room temperature ductility of 12% when aged 500 hr at 1600 °F, and Haynes 25 reaches 6% when aged 1000 hr at 1600 °F. The elevated temperature ductilities of both weld metals are quite good, even in the aged conditions. Hastelloy X-280 maintains a ductility of 40% or more when aged and tested at all four aging temperatures, while Haynes 25 maintains a ductility of 18%.

The aging of Hastelloy X weld metal and Hastelloy X-280 base metal produces a carbide precipitate, probably a mixture of $M_{23}C_6$ and M_6C , in the grain boundaries. The aging of Hastelloy W weld metal at 1000 and 1200 °F creates a carbide precipitate similar to the one produced in Hastelloy X weld metal. Aging Hastelloy W at 1400 and 1600 °F produces a Widmanstätten precipitate network of the beta and gamma phases of the nickel molybdenum binary diagram. Although Hastelloy W does have 5% chromium that suppresses the precipitation of the beta and gamma phases, the prolonged aging of 500 and 1000 hr does cause precipitation of these two phases.

The most severe age embrittling temperatures are 1400 °F for Hastelloy W weld metal and 1200 °F for Hastelloy X weld metal.

BACKGROUND

A brief summary of the physical metallurgy of Hastelloy X, Hastelloy X-280, Hastelloy W and Haynes 25 should help in understanding the effects of aging on the structure of these metals. In general, these four superalloys are solid-solution strengthened alloys, but their excellent strength properties at temperatures above 1200 °F depend on the combined effects of solid-solution strengthening, precipitation-hardening and dispersion hardening. Haynes 25 is a cobalt-base alloy, and the other three alloys are nickel-base alloys.

The plastic deformation of a metal depends on the movement of dislocations through a crystal lattice. Any mechanism that retards the dislocation movement will be a strengthening mechanism for the metal. Solid-solution strengthening is the result of lattice distortions caused by the solution of dissimilar elements in the crystal lattice. The maximum solid-solution strengthening is obtained when a large number of elements are dissolved to form an extremely complex solid solution. These solid solutions unusually have a decreasing solubility of solute in solvent with decreasing temperature. This solvus line effect results in the precipitation of secondary phases at temperatures below the solution temperature.

The amount of secondary phase precipitated is dependent on the temperature, time at temperature, prior thermal mechanical treatment, and the particular alloying elements. The secondary phases of Hastelloy X and Hastelloy X-280 are predominately the carbides $M_{23}C_6$ and M_6C .^{*} Although the secondary phases of Hastelloy W are the carbides $M_{23}C_6$ and M_6C , they can also include the beta and gamma phases of the nickel molybdenum binary diagram.⁽⁶⁻¹⁰⁾ The principal precipitating second phase in Haynes 25 is a Laves phase. The chemical compositions of the materials studied in this investigation are listed in Table I.

^{*}M represents metal or mixture of metals.

TABLE I. Chemical Composition of Weld Specimen Materials

Material	Condition and Form	Weight Percent											
		C	Mn	P	S	Si	Cr	Ni	Co	Fe	Mo	W	V
Hastelloy X	as-drawn welding wire	0.07	0.52	0.018	0.009	0.88	21.57	Bal	1.22	18.09	9.05	0.66	---
Hastelloy W	as-drawn welding wire	0.02	0.59	0.012	0.013	0.42	4.86	Bal	0.56	6.00	23.96	---	0.35
Hastelloy X-280	solution treated plate	0.12	0.45	0.012	0.006	0.42	22	Bal	0.28	17.43	8.84	0.30	---
Haynes 25	solution treated plate	0.07	1.39	0.012	0.009	0.59	19.76	10.19	Bal	1.80	---	14.93	---
AISI 316 Stainless Steel	annealed plate	0.08	2.00	---	---	1.00	17	12	---	Bal	2.5	---	---

In most cases the carbides have a relatively minor effect on strength, but they can have a marked influence on the ultimate ductility. The effect of the carbides on the strength and ductility will depend on the location of the carbides and the rate of loading. The precipitation of $M_{23}C_6$ carbides in the grain boundaries of an alloy can increase the materials elongation at fracture in a creep-rupture test, and decrease the elongation at fracture in a tensile test; however, there is some disagreement as to what constitutes the correct positioning of the carbides.⁽³⁾ The preferred locations of these carbides depends on the prior thermal-mechanical treatment of the materials.

A second complicating factor is the form the carbide takes during precipitation. The carbide $M_{23}C_6$ can precipitate as films, globules, platelets, lamellae, and a cellular growth. The least undesirable grain boundary precipitate form is globular because this form results in a shortening of the microcrack length. Because the M_6C carbide is usually within the grains, it is associated with strengthening of the metal and is not generally considered a contributor to the embrittlement or loss in ductility of a metal. The principal constituent of the metallic radical in $M_{23}C_6$ is chromium; but nickel, cobalt, iron, molybdenum, and tungsten can substitute for chromium. The amount of substitution depends on the availability of the substituting elements and time at temperature. The principal constituents of the metallic radical in M_6C are the heavy elements molybdenum and tungsten, although there is some solubility

for chromium, nickel, and cobalt. Often M_6C is found randomly distributed in the as-cast structure, and $M_{23}C_6$ is usually found in the grain boundaries. The beta and gamma phases in nickel-molybdenum alloys have been shown to be angular particles that tend to form Widmanstatten precipitate networks.⁽⁶⁾ The addition of chromium to these alloys tends to suppress the formation of these beta and gamma phases.⁽¹⁰⁾

In Haynes 25, the Laves phase has been identified as a hexagonal type structure with the probable formula of $(Co, Ni)_2(Cr, W)$ and is usually referred to as Co_2W .⁽²⁾ The preferred locations for this Laves phase in cold-worked and annealed Haynes 25 plate would be in the grains and on the twin planes. The Laves phase Co_2W is the principal embrittling agent in Haynes 25.⁽⁵⁾

TEST RESULTS*

BEND TESTS

The room temperature bend test results are graphically represented in Figure 1 and listed in Table II. In all cases, the failure of a bend specimen

FIGURE 1. The Maximum Elongation in the Outer Fibers of Unaged and Aged Weld Specimens, Bent to Failure in a Longitudinal-Weld Bend Test

*The test specimens and procedures are described in the experimental procedures section of this report.

originated in the weld metal. Therefore, the bend test results represent the room temperature outer fiber maximum elongation of Hastelloy X and Hastelloy W weld metals.

TABLE II. Room Temperature Bend Ductility of Dissimilar Metal Welded Joints, with Hastelloy X and Hastelloy W Weld Metal

Materials*			Aging Conditions		Bend Angle, Degrees	Maximum Weld Elongation, %
Base	Weld	Base	Time, hr	Temperature, °F		
316	X	X-280		unaged	180	39.5
316	W	X-280		unaged	180	29.8
316	X	X-280	500	1000	92	17.5
316	W	X-280	500	1000	99	17.5
316	X	X-280	500	1200	13	3
316	W	X-280	500	1200	58	17.3
316	X	X-280	1000	1000	123	22.1
316	W	X-280	1000	1000	71	17.5
316	X	X-280	1000	1200	121	19
316	W	X-280	1000	1200	128	15.5
25	X	X-280		unaged	180	40
25	W	X-280		unaged	180	30.2
25	X	X-280	500	1400	10	4.5
25	W	X-280	500	1400	14	4.5
25	X	X-280	500	1600	20	7.5
25	W	X-280	500	1600	18	7.5
25	X	X-280	1000	1400	17	7.5
25	W	X-280	1000	1400	18	4.5
25	X	X-280	1000	1600	22	6
25	W	X-280	1000	1600	18	4.5

* 316 denotes AISI 316 Stainless Steel

25 denotes Haynes 25

X-280 denotes Hastelloy X-280

X denotes Hastelloy X

W denotes Hastelloy W

Three typical bend specimen failures of high, moderate and low ductilities are shown in Figures 2, 3, and 4. Joints of 316 SS to Hastelloy X-280 that failed with low ductility differed from the example shown in Figure 4. Due to the high fracture toughness of 316 SS the crack originating in the weld metal did not significantly penetrate the 316 SS base metal. However, the appearance and behavior of the crack in the weld metal and Hastelloy X-280 base metal of these low ductility 316 SS joints was the same as for the low ductility joints of Hastelloy X-280 to Haynes 25. The macro- and microfracture appearances of a typical low ductility failure are shown in Figure 5. There was no apparent difference in the fracture appearance of Hastelloy X and Hastelloy W weld metals in these bend tests. The bend specimen failure shown in Figure 5 represents intergranular fracture in the weld metal and both base metals. When the low ductility bend specimens failed, the first appearance of a crack on the tension surface of the specimen was nearly coincident with complete fracture of the base metal and weld metal.

Both weld metals suffered a loss in room temperature ductility with aging. Although the aging is more sluggish in Hastelloy W than in Hastelloy X, both weld metals reached similar minimum ductilities of 3 to 5%. The only significant differences in room temperature bend ductility of the weld metals is for the unaged condition and the 500 hr age at 1200 °F condition. The unaged bend specimens indicate that Hastelloy X is more ductile than Hastelloy W. Aging at 1200 °F does not significantly affect the ductility of Hastelloy W, but it drastically affects the ductility of Hastelloy X. Hastelloy X reaches its minimum bend ductility when aged at 1200 °F for 500 hr; aging for 1000 hr at 1200 °F causes overaging which results in an increase in bend ductility. Hastelloy W reaches its minimum bend ductility when aged 500 hr at 1400 °F, and when aged 1000 hr at 1600 °F.

TENSILE TESTS

Uniaxial tensile tests of aged and unaged Hastelloy X weld metal, Hastelloy W weld metal, Hastelloy X-280 base metal and Haynes 25 base metal were run at room temperature and at the aging temperatures of 1000, 1200, 1400, and 1600 °F. The results of these tests on the weld metals and base metals are listed in Tables III and IV, respectively.

Side View
 $\frac{7}{8}$ X

Face View
 $\frac{7}{8}$ X

FIGURE 2. Typical Unaged Bend Specimen Failure, High Ductility

Side View
 $\frac{7}{8}X$

Face View
 $\frac{7}{8}X$

FIGURE 3. Typical Aged Bend Specimen Failure, Moderate Ductility

Side View
 $\frac{7}{8}$ X

Face View
 $\frac{7}{8}$ X

FIGURE 4. Typical Aged Bend Specimen Failure, Low Ductility

Fracture Surface, 2 X

Haynes 25, 100 X

Hastelloy X, 100X

FIGURE 5. Typical Fracture Appearance of Low Ductility Bend Test Failure. (The fracture surface is normal to the plane of the 100 X micrographs.)

Fracture Surface, 2 X

Hastelloy X, 100X

Hastelloy X-280, 100 X

ity Bend Test Failure. (The fracture surface is normal to the plane of the 100 X micrographs.)

TABLE III. Mechanical Properties of Hastelloy X and Hastelloy W Weld Metal

Aging Conditions		Material ^(a)	Test Temperature, °F	Strength 10 ³ psi		Total Elongation %
Time, hr	Temperature, °F			Yield* Strength	Ultimate Tensile Strength	
	unaged	X	Room	57.2	82.3	11.3
	unaged	W	Room	64.8	113.5	32.2
	unaged	X	1000	45.9	69.4	24.9
	unaged	W	1000	50.4	88.5	26.2
	unaged	X	1200	40.8	72.9	29.9
	unaged	W	1200	48.8	81.1	28.4
	unaged	X	1400	38.3	53.2	32.7
	unaged	W	1400	43.4	67.7	15.8
	unaged	X	1600	24.9	32.3	29.1
	unaged	W	1600	32.8	48.3	16.8
500	1000	X	Room	59.5	108.5	23.9
500	1000	W	Room	89.3	145.3	17.8
500	1000	X	1000	46.8	73.9	15.7
500	1000	W	1000	76.4	118.4	17.6
500	1200	X	Room	84.6	114.5	1.5
500	1200	W	Room	78.6	142.4	12.3
500	1200	X	1200	62.3	87.1	5.2
500	1200	W	1200	67.1	108.2	12.7
500	1400	X	Room	77.1	126.3	3.7
500	1400	W	Room	91.7	143.7	2.1
500	1400	X	1400	42.3	60.5	18.1
500	1400	W	1400	59.8	88.4	9.6
500	1600	X	Room	59	106.9	4.9
500	1600	W	Room	75.6	138.4	5.0
500	1600	X	1600	21.8	35.5	36.1
500	1600	W	1600	27	43.3	19.1
1000	1000	X	Room	63	85.3	3.5
1000	1000	W	Room	89.8	147.3	18.1
1000	1000	X	1000	48.1	87.4	17.1
1000	1000	W	1000	78.6	124.9	18.0
1000	1200	X	Room	46.3	78.6	8.1
1000	1200	W	Room	67	105	11.3
1000	1200	X	1200	31.3	65.9	15.8
1000	1200	W	1200	47.3	95.3	23.4
1000	1400	X	Room	64.4	131	7.7
1000	1400	W	Room	83.2	149.8	3.7
1000	1400	X	1400	43.4	66.6	32.4
1000	1400	W	1400	28.1	92.2	10.0
1000	1600	X	Room	53.9	113.8	6.2
1000	1600	W	Room	71.8	139.6	5.1
1000	1600	X	1600	24.3	37.5	43.6
1000	1600	W	1600	29.9	47.2	29.6

*This is the 0.2% offset yield strength.

(a) X denotes Hastelloy X
W denotes Hastelloy W

TABLE IV. Mechanical Properties of Hastelloy X-280 and Haynes 25 Base Metal

Aging Conditions		Material ^a	Test Temperature, °F	Strength 10 ³ psi		Total Elongation, %
Time, hr	Temperature, °F			Yield* Strength	Ultimate Tensile Strength	
unaged		X-280	Room	49.1	104	41.7
unaged		25	Room	67	143.3	48.6
unaged		X-280	1000	37.9	84.1	46.1
unaged		X-280	1200	31.8	79.6	48.1
unaged		X-280	1400	37.9	63	55.9
unaged		25	1400	44.6	75.4	18
unaged		X-280	1600	26.7	39	65.9
unaged		25	1600	34.1	51.8	20.1
500	1000	X-280	Room	48.9	107.1	45
500	1000	X-280	1000	37.2	85.1	48.5
500	1200	X-280	Room	52.1	11.6	31.3
500	1200	X-280	1200	40	84.7	40.5
500	1400	X-280	Room	59	128.7	16.6
500	1400	25	Room	77.8	128.9	12.5
500	1400	X-280	1400	40.4	64.3	47.2
500	1400	25	1400	40.4	77	41.6
500	1600	X-280	Room	49.3	109.1	11.6
500	1600	25	Room	82.4	15.5	7.3
500	1600	X-280	1600	27.4	43.9	46.3
500	1600	25	1600	26.9	42.3	52.1
1000	1000	X-280	Room	49.8	111.2	42.3
1000	1000	X-280	1000	35.5	93.8	41.3
1000	1200	X-280	Room	45.9	109	30.8
1000	1200	X-280	1200	30.7	82.5	40
1000	1400	X-280	Room	55.1	128.4	18.6
1000	1400	25	Room	79.3	134.8	7.5
1000	1400	X-280	1400	35.9	65.3	48.1
1000	1400	25	1400	45.6	87.6	16.3
1000	1600	X-280	Room	47.8	111.5	13.7
1000	1600	25	Room	82.3	157.9	5.9
1000	1600	X-280	1600	22.2	36.4	57.2
1000	1600	25	1600	30.7	47.9	35.9

*This is the 0.2% offset yield strength.

^aX-280 denotes Hastelloy X-280
25 denotes Haynes 25

The influence of aging at temperatures between 1000 and 1600 °F on the room temperature mechanical properties of the weld metals is shown in Figure 6. These tensile tests indicate that although the unaged weld metals have nearly the same yield strength, Hastelloy W is considerably more ductile than Hastelloy X. Aging Hastelloy X at 1000 °F for 500 hr significantly improves its ductility and does not appreciably change its yield strength. Additional aging of Hastelloy X for 1000 hr at 1000 °F results in a significant loss in ductility to a low of 3.5%, and aging does not appreciably change the yield strength. Hastelloy W aged at 1000 °F indicates very little difference in the room temperature mechanical properties for material aged for 500 and 1000 hr. This aging of Hastelloy W does increase its yield strength to 89,000 psi and lowers its ductility to nearly 18%.

Aging the two weld metals for 500 hr at 1200 °F raises their yield strengths to approximately 80,000 psi, and it reduces the ductility of Hastelloy X to 1.5% and the ductility of Hastelloy W to 12%. Additional aging of Hastelloy X at 1200 °F results in overaging, with the 1000 hr age causing an improvement of ductility to 8% and a decrease in yield strength to 46,000 psi. Additional aging of Hastelloy W at 1200 °F did not appreciably affect the ductility but lowered the yield strength to 67,000 psi.

Aging Hastelloy X for 500 hr at 1400 °F lowered its ductility to 4% and raised its yield strength to 77,000 psi. Additional aging of Hastelloy X at this temperature results in overaging, an increase in ductility to 8%, and a decrease in yield strength to 64,000 psi. Similar aging of Hastelloy W at 1400 °F results in a decrease in ductility to 2% and an increase in yield strength to 92,000 psi. Additional aging of Hastelloy W at this temperature does not appreciably affect its room temperature ductility but slightly lowers the yield strength which indicates overaging is beginning to occur.

Both of the weld metals aged at 1600 °F for 500 and 1000 hr have room temperature ductilities of approximately 5%. Hastelloy X aged at 1600 °F for 500 and 1000 hr has yield strengths of 59,000 and 54,000 psi, respectively. The room temperature tensile test data indicate that both metals reach similar minimum ductilities for a given maximum embrittlement. The data also indicate that Hastelloy X reaches its minimum ductility much faster than does Hastelloy W.

FIGURE 6. Room Temperature Tensile Test Properties of Unaged and Aged Hastelloy X and Hastelloy W Weld Metals

The effect of aging in the temperature range of 1000 to 1600 °F on the elevated temperature mechanical properties of the weld metals is shown in Figure 7. The elevated temperature mechanical properties of the unaged weld metals appear to follow the trend (indicated by data in the literature) for wrought Hastelloy W and Hastelloy X. Unaged Hastelloy X weld metal maintains a much higher ductility over the temperature range in question than does unaged Hastelloy W weld metal; the ductility of Hastelloy W decreases to 16 to 17% when tested at 1400 and 1600 °F. Unaged Hastelloy X tested between 1000 and 1600 °F maintains a ductility of greater than 25%. Over the same temperature range, the unaged weld metals maintain a relatively high level of yield strength, and there is very little difference between the yield strengths of either weld metal.

Aging of Hastelloy W weld metal at 1000 °F decreases its ductility at 1000 °F to 18%; there is very little difference in the ductility of the material aged 500 and 1000 hr at 1000 °F. This aging of Hastelloy W at 1000 °F also results in an increase in its yield strength at this temperature. Again, there is very little difference between the yield strengths of the material aged 500 hr and the material aged 1000 hr. Aging Hastelloy X at 1000 °F for 500 and 1000 hr decreases its ductility to 16%, but its yield strength is not significantly affected at that temperature.

Aging Hastelloy W and Hastelloy X at 1200 °F significantly changes their 1200 °F mechanical properties. Aging 500 hr lowers the ductility of Hastelloy W to 13% and the ductility of Hastelloy X to 5%. Continued aging for 1000 hr results in overaging, with the ductilities of Hastelloy W and Hastelloy X raised to 23 and 16%, respectively. This aging of both weld metals at 1200 °F results in the yield strengths for the 500 hr age being increased to values greater than 60,000 psi. The overaging caused by the 1000 hr age lowers the yield strengths of both weld metals.

Aging of Hastelloy W for 500 hr at 1400 °F results in a decrease in the ductility at that temperature to 10% and an increase in yield strength at that temperature to 60,000 psi. Continued aging of Hastelloy W at 1400 °F for 1000 hr does result in overaging and a decrease in yield strength to 28,000 psi but does not significantly change the 10% ductility at that temperature.

FIGURE 7. Elevated Temperature Tensile Test Properties of Unaged and Aged Hastelloy X and Hastelloy W Weld Metals

Aging Hastelloy X for 500 hr at 1400 °F decreases the ductility at that temperature to 18%, but does not significantly change the yield strength from the value of the unaged metal at that temperature. Continued aging of Hastelloy X at 1400 °F for 1000 hr results in overaging, an increase in ductility at that temperature to 32%, and does not significantly change its yield strength at that temperature.

Aging Hastelloy W for 500 hr at 1600 °F produces a slight recovery in ductility from the low value of 17% of the unaged weld metal at that temperature and a decrease in yield strength to 22,000 psi. Continued aging of Hastelloy W at 1600 °F for 1000 hr results in overaging and an increase of ductility at that temperature to 30% and does not significantly change its yield strength at that temperature. Hastelloy X aged at 1600 °F for 500 hr results in an increase in ductility at that temperature to 36% and a decrease in yield strength at that temperature to 22,000 psi. Continued aging of Hastelloy X for 1000 hr at 1600 °F results in still further increase in its ductility at 1600 °F to 44% and a slight recovery in yield strength at 1600 °F to 24,000 psi.

The room temperature mechanical properties of Hastelloy X-280 base metal and Haynes 25 base metal aged for 500 and 1000 hr at the temperatures of 1000, 1200, 1400, and 1600 °F are shown in Figure 8. As shown in the figure, there is very little difference in the ductility or yield strength of Hastelloy X-280 aged at 500 and 1000 hr at any of these aging temperatures. There is also very little difference in the room temperature ductility or yield strength of Haynes 25 aged at 1400 and 1600 °F for 500 or 1000 hr. Although the figures do indicate that both of these base metals suffer from some age embrittlement in this temperature range the minimum ductilities are approximately 10%.

The elevated temperature mechanical properties of Hastelloy X-280 base metal and Haynes 25 base metal are shown in Figure 9. As indicated in the figure aging 500 and 1000 hr at the indicated aging temperatures does not significantly change the mechanical properties of Hastelloy X-280 at these test temperatures. There is some effect of age embrittlement on the elevated temperature mechanical properties of Haynes 25 aged and tested

FIGURE 8. Room Temperature Tensile Test Properties of Unaged and Aged Hastelloy X-280 and Haynes 25 Base Metals

FIGURE 9. Elevated Temperature Tensile Test Properties of Unaged and Aged Hastelloy X-280 and Haynes 25 Base Metals

at 1400 and 1600 °F. Aging Haynes 25 for 500 hr at 1400 and 1600 °F results in an increase in ductility at these temperatures to over 40%. Continued aging for a period of 1000 hr at 1400 and 1600 °F results in embrittlement of Haynes 25 and a decrease in the ductility at 1400 °F to a value close to the unaged materials ductility. It also results in a decrease in the ductility at 1600 °F to a value that is still 80% higher than the unaged materials ductility at that temperature.

HARDNESS STUDIES

The effects of aging for 500 and 1000 hr at 1000 and 1200 °F on the room temperature hardness of welded joints of Hastelloy X-280 to 316 SS with Hastelloy X weld metal are shown in Figure 10. The hardness

FIGURE 10. Room Temperature Hardness Profiles of Unaged and Aged Weld Specimens of Hastelloy X-280 to AISI 316 Stainless Steel with Hastelloy X Weld Metal

traverses indicate the most significant hardening of this welded joint occurs with the 500 hr age at 1200 °F. This aging for 500 hr at 1200 °F causes significant hardening of both Hastelloy X-280 base metal and Hastelloy X weld metal. Aging for 500 hr at 1000 °F does not significantly change the hardness of the base metals or the weld metal. Aging for 1000 hr at 1000 and 1200 °F causes overaging and the hardness of this joint is lowered to a value less than the hardness of the unaged joint.

The effects of aging for 500 and 1000 hr at 1000 and 1200 °F on the room temperature hardness of welded joints of Hastelloy X-280 to 316 SS with Hastelloy W weld metal are shown in Figure 11. These hardness

FIGURE 11. Room Temperature Hardness Profiles of Unaged and Aged Weld Specimens of Hastelloy X-280 to AISI 316 Stainless Steel with Hastelloy W Weld Metal

traverses indicate that there is significant age hardening of the weld metal when aged at all four of the aging conditions. In addition there is considerable

age hardening of the Hastelloy X-280 base metal aged 500 hr at 1200 °F. The hardening of this Hastelloy W welded joint aged 500 hr at 1200 °F is very similar to the hardening of the Hastelloy X welded joint aged 500 hr at 1200 °F. Aging for 1000 hr at 1200 °F causes overaging of this joint, with the Hastelloy X-280 base metal returning to the hardness level of the unaged base metal but the Hastelloy W weld metal only recovers to a point that is still significantly above the hardness of the unaged weld metal. Aging for 500 hr at 1000 °F does not significantly change the hardness of the Hastelloy X-280 base metal but significantly changes the hardness of the Hastelloy W weld metal. Aging for 1000 hr at 1000 °F does not appreciably change the hardness of this joint from the hardness of the same joint aged 500 hr at 1000 °F. Each of these four aging treatments on this welded joint result in a hardness profile that has a sharp transition in hardness going from Hastelloy X-280 base metal to Hastelloy W weld metal and back down to 316 SS base metal. As expected the hardness of the 316 SS base metal does not significantly change with these aging treatments.

The effects of aging 500 and 1000 hr at 1400 and 1600 °F on the room temperature hardness of welded joints of Hastelloy X-280 to Haynes 25 with Hastelloy X weld metal are shown in Figure 12. These hardness traverses indicate that there is significant age hardening in both base metals and Hastelloy X weld metal. It is interesting to note that aging causes more hardening in the Haynes 25 base metal than in either Hastelloy X weld metal or Hastelloy X-280 base metal. In the joints that have been aged 500 hr at 1600 °F, 1000 hr at 1600 °F, and 1000 hr at 1400 °F, there is a sharp transition of hardness in going from Hastelloy X-280 base metal to Haynes 25 base metal. Aging 500 hr at 1400 °F results in significant age hardening in both base metals and weld metal but there is not much difference in hardness across the welded joint. Age hardening is indicated to be slower in the Haynes 25 base metal than the Hastelloy X weld metal but the ultimate hardness of the aged Haynes 25 base metal is considerably higher than the hardness of the aged Hastelloy X weld metal. The unaged joint has a sharp transition in hardness between Hastelloy X weld metal and Haynes 25 base metal.

FIGURE 12. Room Temperature Hardness Profiles of Unaged and Aged Weld Specimens of Hastelloy X-280 to Haynes 25 with Hastelloy X weld Metal

The effects of aging for 500 and 1000 hr at 1400 and 1600 °F on the room temperature hardness of welded joints of Hastelloy X-280 to Haynes 25 with Hastelloy W weld metal are shown in Figure 13. These hardness traverses indicate there is more hardening in the weld metal of this joint than there is in the Hastelloy X weld metal of the similar joint. This hardening of the weld metal results in abrupt hardness transitions across the fusion zones of Hastelloy W weld metal to Haynes 25 and Hastelloy X-280 base metals. Aging 500 hr at 1400 °F results in the most severe hardening of Hastelloy X-280 base metal and Hastelloy W weld metal. Aging for 1000 hr at 1400 °F causes overaging of Hastelloy X-280 and Hastelloy W but there still is a sharp transition in hardness across the fusion zones between the weld metal and the two base metals. Aging 500 hr

at 1600 °F produces the most significant hardening of Haynes 25. Aging for 1000 hr at 1600 °F does not significantly change the hardness profile of this welded joint from the hardness of the same welded joint aged 500 hr at 1600 °F. All of these hardness traverses indicate that the aging results in abrupt transitions in hardness at both weld fusion zones. In addition, the unaged joint has a high hardness at the fusion zone between Hastelloy W and Haynes 25.

FIGURE 13. Room Temperature Hardness Profiles of Unaged and Aged Weld Specimens of Hastelloy X-280 to Haynes 25 with Hastelloy W Weld Metal

METALLOGRAPHIC EXAMINATIONS

The microstructure of Hastelloy W weld metal in the unaged condition is shown in Figure 14. The unaged microstructure has a significant amount of inclusions and carbides probably M_6C , randomly distributed within the grains. The effects of aging 500 and 1000 hr at 1000, 1200,

Hastelloy W

Hastelloy X

FIGURE 14. Unaged Hastelloy X and Hastelloy W Weld Metal, 500 X

1400, and 1600 °F are shown in Figures 15 and 16. Aging at 1000 °F causes significant second phase precipitation at the grain boundaries and between the dendrites. There is not much difference between the microstructures of this weld metal aged 500 hr at 1000 °F and aged 500 hr at 1200 °F, and between the microstructures of this metal aged 1000 hr at 1000 °F and aged 1000 hr at 1200 °F. The 1000 hr exposure at 1200 °F results in a very complicated carbide network within the grains. Aging for 500 hr at 1400 °F produces a Widmanstatten-type precipitate concentrated near the grain boundaries. Continued aging for 1000 hr at 1400 °F causes some of the Widmanstatten precipitate to agglomerate but the Widmanstatten structure is still present. This precipitate produced at 1400 °F corresponds to the beta phase of the nickel-molybdenum binary diagram. Aging at 1600 °F also produces a Widmanstatten precipitate but the precipitate network for 500 hr at 1600 °F is much coarser than the network for 500 hr at 1400 °F. Aging 1000 hr at 1600 °F causes the precipitate to agglomerate to a point where the Widmanstatten precipitate network is no longer present. This precipitate produced at 1600 °F corresponds to the gamma phase of the nickel-molybdenum binary diagram.

The microstructure of Hastelloy X weld metal in the unaged condition is shown in Figure 14. The effects of aging for 500 and 1000 hr at 1000, 1200, 1400, and 1600 °F are shown in Figures 17 and 18. Aging for 500 hr at 1000 °F does not appreciably change the appearance of the microstructure. The unaged and 500 hr at 1000 °F aged microstructures indicate a significant amount of inclusions and carbides, probably M_6C , randomly distributed within the grains.⁽¹¹⁾ The second phases precipitated at all four aging temperatures are probably a mixture of M_6C and $M_{23}C_6$, with the majority of the precipitate in the M_6C form.⁽⁸⁾ Aging for 1000 hr at 1000 °F produces a significant amount of carbide precipitate in the grain boundaries and between the dendrites. Aging 500 hr at 1200 °F produces a large amount of a very fine carbide precipitate between the dendrites. Continued aging for 1000 hr at 1200 °F overages the structure and produces carbide precipitate clusters around large carbides within the grains and relatively massive carbides in the grain boundaries. Aging at 1400 °F produces a very fine precipitate randomly distributed within the grains.

FIGURE 15. Hastelloy W Weld Metal Aged 500 and 1000 hr at 1000 and 1200 °F, 500 X

FIGURE 16. Hastelloy W Weld Metal Aged 500 and 1000 hr at 1400 and 1600 °F, 500 X

FIGURE 17. Hastelloy X Weld Metal Aged 500 and 1000 hr at 1000 °F, 500 X

FIGURE 18. Hastelloy X Weld Metal Aged 500 and 1000 hr at 1400 and 1600 °F, 500 X

There appears to be more precipitate produced, at 1400 °F, by the 1000 hr age than the 500 hr age but the particles are still hard to resolve at 500 X. Aging at 1600 °F produces relatively large precipitate particles randomly distributed within the grains as well as in the grain boundaries. The 1000 hr age at 1600 °F results in agglomeration of the particles produced by the 500 hr age at 1600 °F.

The microstructure of Hastelloy X-280 plate in the solution treated condition is shown in Figure 19. The effects of aging for 500 and 1000 hr at 1000, 1200, 1400, and 1600 °F on the microstructure of Hastelloy X-280 plate are shown in Figures 20 and 21. The second phases precipitated at all four aging temperatures are probably a mixture of M_6C and $M_{23}C_6$, with the majority of the precipitate in the M_6C form.⁽⁸⁾ Aging at 1000 °F produces a continuous carbide network in the grain boundaries and heterogeneously scattered within the grains. Aging 500 hr at 1200 °F produces a thick continuous carbide network in the grain boundaries and a fine carbide precipitate on the twin planes. Continued aging of 1000 hr at 1200 °F causes considerable agglomeration of the grain boundary carbides and the development of a random distribution of carbides within the grains. Aging at 1400 °F produces a large amount of carbide precipitate in the grain boundaries and randomly distributed within the grains. Aging at 1600 °F also produces a large amount of carbide precipitate in the grain boundaries and within the grains, but there is less precipitate developed at 1600 °F than there was at 1400 °F. At 1400 and 1600 °F there is not an appreciable difference between the microstructures produced by aging 500 hr and by aging 1000 hr.

The microstructure of Haynes 25 plate in the solution treated condition is shown in Figure 19. The effects of aging 500 and 1000 hr at 1400 and 1600 °F are shown in Figure 22. Aging 500 hr at 1400 °F produces a precipitate, probably the Laves phase, network in the grain boundaries. Continued aging of 1000 hr at 1400 °F produces an extensive precipitate network on the twin planes. Aging 500 hr at 1600 °F produces a microstructure similar to the 1000 hr at 1400 °F aged microstructure except the precipitate particles are larger in the 1600 °F microstructure than the 1400 °F microstructure. There is no significant difference between the 1600 °F microstructures aged 500 and 1000 hr.

Hastelloy X-280

Haynes 25

FIGURE 19. Unaged Hastelloy X-280 and Haynes 25 Base Metals, 500 X

FIGURE 20. Hastelloy X-280 Base Metal Aged 500 and 1000 hr at 1000 °F, 500 X

FIGURE 21. Hastelloy X-280 Base Metal Aged 500 and 1000 hr at 1400 and 1600 °F, 500 X

FIGURE 22. Haynes 25 Base Metal Aged 500 and 1000 hr at 1400 and 1600 °F, 500 X

EXPERIMENTAL

The weld specimens used in this program were intended to simulate field-welded, thick-wall, dissimilar-metal pipe joints. The major steps in the testing program are schematically illustrated in Figure 23. The materials used include Hastelloy X and Hastelloy W weld filler wires, and Haynes 25, 316 SS, and Hastelloy X-280 plates. The chemical analysis of these materials are listed in Table I. The weld specimens represent dissimilar metal joints of Hastelloy X-280 to Haynes 25 and Hastelloy X-280 to 316 SS, welded with either Hastelloy X or Hastelloy W filler wire.

FIGURE 23. The Major Steps in the Testing Program

The Hastelloy X-280 and Haynes 25 plates were solution heat treated, according to the manufacturer's recommendations, before specimens were prepared for welding. The base metal sections of the weld specimens were made from 0.56 in. thick plate and machined to the dimensions shown in Figure 24a. Two base metal sections were joined, by the tungste-electrode inert gas (TIG) welding process, to form weld specimens of the dimensions shown in Figure 24b. A total of 60 weld specimens were made for this program. Hastelloy X and Hastelloy W weld filler wires $\frac{3}{32}$ in. in diameter were used. The welding procedure minimized heating to the extent consistent with full penetration, and the specimens were cooled to room temperature between passes. Each specimen required six 9.5 in. long welding passes.* During welding, the base metal sections were tightly clamped so that the finished weld specimen was reasonably flat. Extreme warping of the weld specimen would have occurred without this clamping. The residual stresses caused by the welding and the clamping should be representative of field welded joints in the ATR gas loop. Each weld specimen was radiographically inspected according to Section VIII of the ASME boiler and pressure vessel code for unfired pressure vessels. All but 2 of the 60 weld specimens successfully passed the code inspection. The two specimens that did not meet the specifications failed due to lack of penetration and/or lack of fusion. All 60 specimens were included in the testing program, and subsequent destructive tests were carefully watched to see if failure occurred at the defects indicated by the radiographic inspection. In all cases specimen failure was not caused by defects that failed to meet the radiographic inspection specifications. After the inspection, the weld face side of the specimens was machined flat to a number 16 or better finish.

The weld specimens were aged in an inert atmosphere at temperatures of 1000, 1200, 1400, and 1600 °F. The specimens were placed in metal boxes and the boxes were welded shut. Each box was equipped with three thermocouple entry ports, a gas inlet, and a gas outlet. The boxes were placed inside a furnace, specially built for this program, and the furnace temperature was controlled by a thermocouple measuring the temperature in the

*The welding was done by Technical Welding Specialists.

FIGURE 24. Dimensions of Base Metal Sections and Longitudinal Weld Bend-Specimens

center of the box. Two other thermocouples, measuring temperature at opposite ends of the box, were used to periodically check the temperature gradient in the box. The furnaces were completely surrounded by 4 in. of insulation, and the temperature at any point in the box was within $\pm 10^{\circ}\text{F}$ of the control thermocouple temperature. The inert atmosphere in each of the boxes was maintained by a slow purge of argon. Although bottled argon does have a significant impurity content, the weld specimens maintained a relatively clean surface finish throughout the aging. To remove specimens from the furnace, the boxes were cut open, the specimens removed, and the boxes resealed by welding.

The aged weld specimens were bend tested to determine the relative ductility of the welded joints. The bend test was a longitudinal-weld free bend of the entire weld specimen, with the weld-face being deformed in tension (Figure 23). A grid of 0.25 in. squares was inked on the weld-face side of the specimens, and this grid was used to measure the outer fiber elongation of the specimens after being bent to failure. Failure was designated as the point where an open defect of 0.125 in. or greater, in any dimension, occurred on the weld-face of the bend specimen. The weld specimens were given an initial bend of 10 to 15 degrees and then a free bend to failure. The initial bend and free bend devices are illustrated in Figure 25. The elongation of a joint or weld specimen was measured as the maximum elongation in 0.25 in. of the outer fibers of the material where failure originated. This bend test is a qualitative measure of the welded joints ductility.

Uniaxial tensile specimens of all base-metal and all weld-metal were made from the ends of the bend specimens. The tensile specimen dimensions are shown in Figure 26. It was possible to obtain two all weld-metal and eight all base-metal tensile specimens from each bend specimen. The material used to make these tensile specimens was relatively undisturbed by the bend test. The tensile specimens were tested on a Instron testing machine at a cross-head speed of 0.005 in./min. The base-metal tensile specimens that were tensile tested did not contain any weld heat-affected-zone material.

FIGURE 25. Schematic Illustration of Bend Test

Tensile Specimen

FIGURE 26. Dimensions of Tensile Test Specimens, in.

Hardness profiles of the welded joints were also obtained from samples of the bend specimens. The relative orientation of these hardness profiles is indicated in Figure 23. The hardness was measured on a standard Rockwell hardness tester, and the hardness numbers were converted to Vickers hardness numbers for representation in this report.

All data points presented in this report are the average of tests on two weld specimens. The bend test ductility was determined by two separate bend tests on two weld specimens of the same joint materials and aging treatment. Two tensile specimens, each obtained from a different weld specimen, were used to determine each of the tensile test data points. Two hardness profile specimens, each obtained from a different weld specimen, were used to determine the hardness profile for a given joint material combination and aging treatment.

The preparation of samples for metallographic examination was difficult. The difficulty was caused by the difference in etching behavior of the aged material. In general each material required a different etchant for each of the material's aging treatments.

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