ORNL
MASTER COPY

OAK RIDGE NATIONAL LABORATORY
operated by
UNION CARBIDE CORPORATION
NUCLEAR DIVISION
for the
U.S. ATOMIC ENERGY COMMISSION

ORNL - TM - 2483

16

MAR 26 1969

PREIRRADIATION AND POSTIRRADIATION MECHANICAL PROPERTIES OF HASTELLOY N WELDS

H. E. McCoy
D. A. Canonico

Full public release
Dave Hamrin
6/28/2013

[redacted]

ORNL-TM-2483

Contract No. W-7405-eng-26

METALS AND CERAMICS DIVISION

PREIRRADIATION AND POSTIRRADIATION MECHANICAL PROPERTIES OF HASTELLOY N WELDS

H. E. McCoy and D. A. Canonico

Submitted to the Welding Journal without the Appendix.

MARCH 1969

OAK RIDGE NATIONAL LABORATORY
Oak Ridge, Tennessee
operated by
UNION CARBIDE CORPORATION
for the
U.S. ATOMIC ENERGY COMMISSION

CONTENTS

PREIRRADIATION AND POSTIRRADIATION MECHANICAL PROPERTIES OF HASTELLOY N WELDS

H. E. McCoy and D. A. Canonico

ABSTRACT

Welds were made by the TIG process in several heats of Hastelloy N. The mechanical properties of transverse weld samples and the base metal were compared in tensile tests over the range of 75 to 1600°F and in creep tests at 1200°F. The as-fabricated welds exhibited lower fracture strains than the base metal under all test conditions, but the properties of the welds were improved markedly by post-weld heat treatments. The postirradiation tensile and creep properties of the welds and base metal at elevated temperatures were about the same, although the properties were widely different before irradiation.

INTRODUCTION

Hastelloy N is a trade name given the solid-solution-strengthened nickel-base alloy developed at the Oak Ridge National Laboratory specifically for use in molten fluoride salts up to 1500°F (ref. 1). The alloy was originally designated INOR-8 and has a nominal composition of Ni–17% Mo–7% Cr–5% Fe. This material is the primary metallic structural material in the Molten Salt Reactor Experiment (MSRE) which achieved criticality on June 1, 1965, at Oak Ridge, Tennessee.[2] This alloy has

been used for numerous other applications, and we anticipate that a slightly modified composition will be used in a future molten salt breeder reactor experiment.

The weldability of this material has received considerable attention in our program.[3,4] We found it necessary to establish a weldability test to determine whether specific heats of material could be joined satisfactorily. Through this screening process, it was possible to select heats for use in fabricating the MSRE components.

During the past few years we have found that Hastelloy N as well as other nickel- and iron-base alloys are subject to a type of neutron irradiation damage that decreases the creep-rupture strength and fracture strain at elevated temperatures.[5,6] This damage is rather general for all austenitic iron- and nickel-base alloys and is attributed to the production of helium within the material due to the interaction of a neutron of thermal energy with ^{10}B to produce ^{7}Li and ^{4}He (refs. 7–14). The property changes due to the helium must be evaluated for base metal and welds.

In the present paper, we shall show how the mechanical properties of transverse welds differ from those of the base metal and what types of annealing treatments can be used to improve the properties of welds. We shall also show how neutron irradiation alters the mechanical properties of the welds and the base metal. The mechanical properties measured in this study were tensile properties over the temperature range of 75 to 1600°F and creep properties at 1200°F.

EXPERIMENTAL DETAILS

Several heats of Hastelloy N were used in this study and their chemical compositions and other pertinent details are given in Table 1. Heats 65-552 and 2477 were vacuum melted; all others were air melted. All base metal was annealed for 1 hr at 2150°F prior to testing unless otherwise specified.

Seven weldments were involved in this study and the details are given in Table 2. The welds were all highly restrained and made by the manual tungsten-arc welding process. A standard welding procedure was used on all the welds, although some deviation was necessary in welds 7 and 8 because of the type of filler metal being used. The joint configuration and pass sequence are shown in Fig. 1. The primary working direction in the plates was perpendicular to the weld axis.

We used small mechanical property samples throughout the study, since this geometry was required for the irradiation experiments (Fig. 1). The samples were cut perpendicular to the weld axis and parallel to the stringers. Three layers of specimens were cut from all the welds except No. 3; we could find no systematic variation in properties from top to bottom. However, the gage portions of the transverse weld specimens did contain different amounts of weld metal since the joint tapered (Fig. 1).

The tensile tests were run in an Instron Universal testing machine (10,000-lb capacity) at strain rates of 0.05 or 0.002 min^{-1}. The strain measurements were taken from the crosshead travel. A similar machine was located in a hot cell for testing the irradiated samples. The laboratory creep-rupture tests were run in standard lever-arm creep machines. The strain measurements were taken from the pull-rod displacement after

Table 1. Chemical Analysis of Test Materials

Heat Designation	Melting Practice	Composition (wt %) Cr	Fe	Mo	C	Si	Co	W	Mn	V	P	S	Al	Ti	Cu	(ppm) B	O	N
5065	Air	7.2	3.9	16.5	0.065	0.60	0.08	0.04	0.55	0.22	0.004	0.007	0.01	0.01	0.01	10–37	16	110
5067	Air	7.4	4.0	17.3	0.060	0.43	0.08	0.06	0.50	0.30	0.013	0.007	0.01	0.01	0.015	10–24	12	100
5101	Air	6.9	3.9	16.4	0.05	0.63		0.05	0.44	0.34	0.001	0.009	— 0.02[a] —			3.5		
5055	Air	7.9	3.8	16.2	0.06	0.61		0.03	0.69	0.21	0.006	0.008	0.06	0.02		50		
2477	Vacuum	7.05	4.25	16.3	0.057	0.015	0.14	0.47	0.04		0.008	0.003	0.055	0.10	0.10	8	5	7
65-552	Vacuum	6.89	4.06	16.2	0.045	0.16	0.050	0.006	0.45	< 0.0005	0.002	0.006	0.25			0.5	1.3	9

[a] Combined amount of aluminum and titanium.

Table 2. Identification of Welds Made in This Study

Weld Number	Base Metal	Filler Metal	Thickness of Weld (in.)
1	5065, 5067	5101	1
3	2477	2477	1/2
4	5065, 5067	5101	1
5	5065, 5067	5055	1
6	5065	65-552	1 1/8
7	5065	5055 + Al_2O_3 [a]	1 1/8
8	5065	5055 + WC [a]	1 1/8

[a] Filler material was 1/8 in. wire of heat 5055 plasma sprayed with 0.002 in. of indicated material.

the load was applied. The postirradiation creep-rupture tests were run in lever-arm creep machines located in the ORNL hot cells. The strain was measured by an extensometer with rods attached to the upper and lower specimen grips. The relative movement of these two rods was measured by a linear differential transformer. All tests were run in an air environment. We used a standard equilibrating time of 1/2 hr for each test prior to loading.

The irradiations were conducted in two facilities: the ETR, Idaho Falls, Idaho, and the ORR, Oak Ridge, Tennessee. A core facility was used in the ETR where the thermal and fast (> 1 Mev) fluxes were each 3.2×10^{14} neutrons cm^{-2} sec^{-1}. The fluence obtained there was 5×10^{20} neutrons/cm^2. The ETR experiments were uninstrumented and the design temperatures were either less than 300°F or 1112 ± 180°F. A core facility was also used in the ORR where the thermal flux was

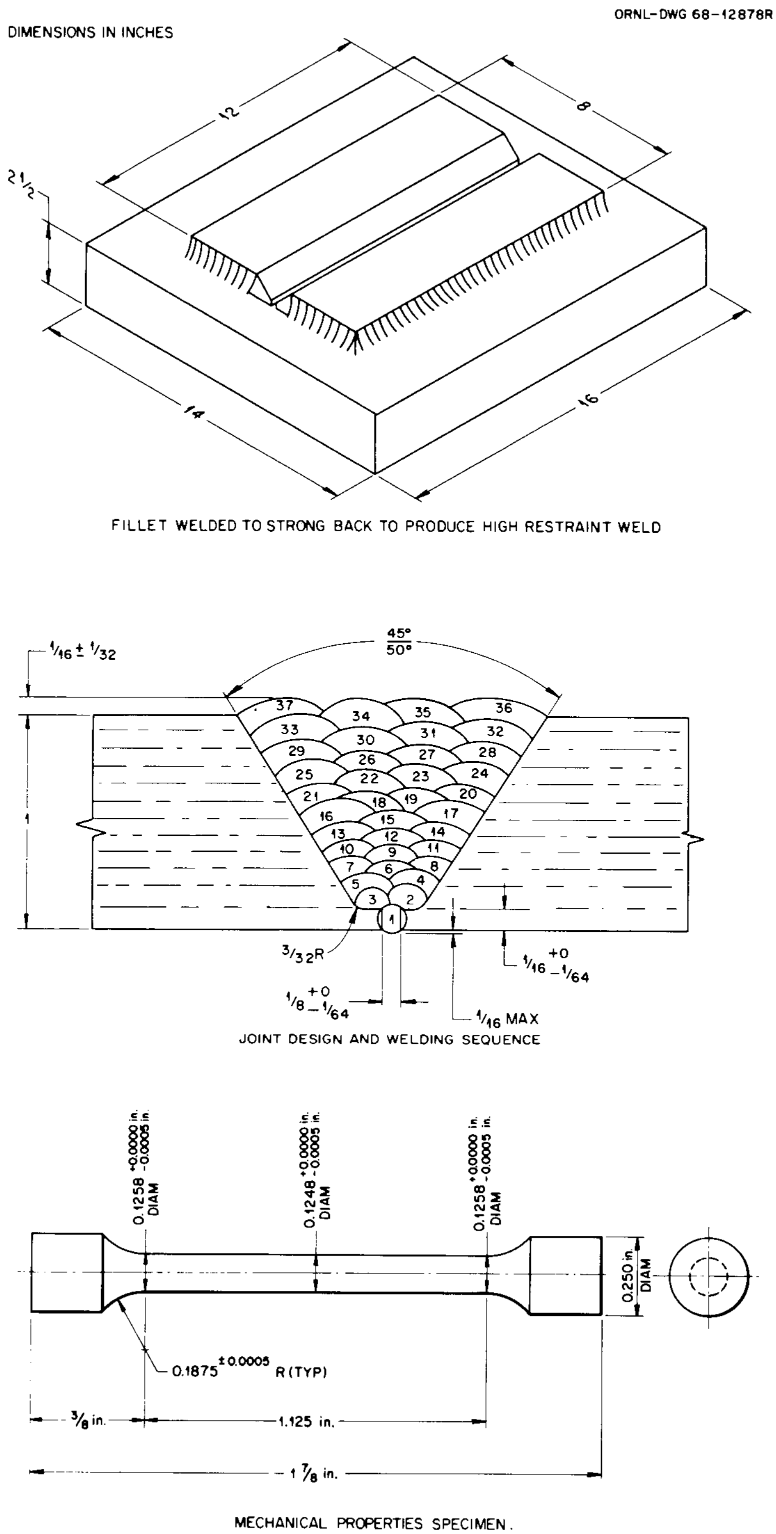


Fig. 1. The INOR-8 High-Restraint Weldability Test Specimen Used to Provide Samples for the Mechanical Properties Study.

2.5×10^{14} neutrons cm^{-2} sec^{-1}. The fluences obtained were 8.5×10^{20} neutrons/cm^2 thermal and 7.0×10^{20} neutrons/cm^2 fast (> 1 Mev). The temperature in this facility was 110°F. We could not observe any effect of fluence over the small range involved here, so the data are presented without reference to a particular experiment. Irradiation temperature was important in some cases and will be indicated.

EXPERIMENTAL RESULTS

Tensile Properties*

The variation of the fracture strain in a tensile test with test temperature is shown in Fig. 2 for both transverse welds and base metal. The behavior of this particular heat of base metal is typical for Hastelloy N. The decrease in fracture strain above 1100°F is due to

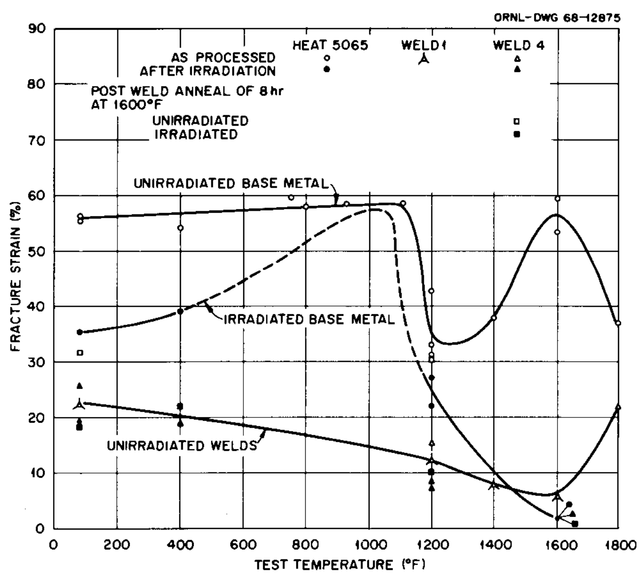


Fig. 2. Ductility of Hastelloy N Welds and Base Metal in Tensile Tests at a Strain Rate of 0.05 min^{-1}.

*See Tables A-1 and A-2, Appendix, for the tabulated tensile data.

the transition from transgranular to intergranular fracture. The recovery above 1400°F is associated with increasing grain boundary mobility. The fracture strain of the welds is much lower. Most of the welds fractured in the weld metal, but several samples tested at 75 and 392°F did fail in the base metal.

The fracture strain after irradiation is also shown in Fig. 2. Since these samples were irradiated at less than 300°F, there is some displacement damage in the base metal and the fracture strain is lower at test temperatures of 75 and 392°F. At test temperatures of 1200 and 1600°F, where intergranular fracture predominates, the fracture strain of the base metal is reduced dramatically. The postirradiation fracture ductility of the irradiated welds is also shown in Fig. 2. The strain at fracture of the welds is reduced slightly by irradiation, but the changes are much less than those observed for the base metal. Thus, although the ductility of the welds is much lower than that of the base metal before irradiation, the ductilities of welds and base metal at elevated temperatures are quite similar after irradiation.

The yield strengths of typical welds and base metal are compared in Fig. 3. The yield strengths of the transverse weld samples are consistently higher than those of the base metal. This same observation was also made by Gilliland and Venard.[3] After a postweld anneal of 8 hr at 1600°F, the welded and base metal samples have very similar yield strengths.

Irradiation increases the yield strength of the base metal and welded samples at low temperatures. This damage anneals out as the test temperature is increased, and the welds and base metal show similar

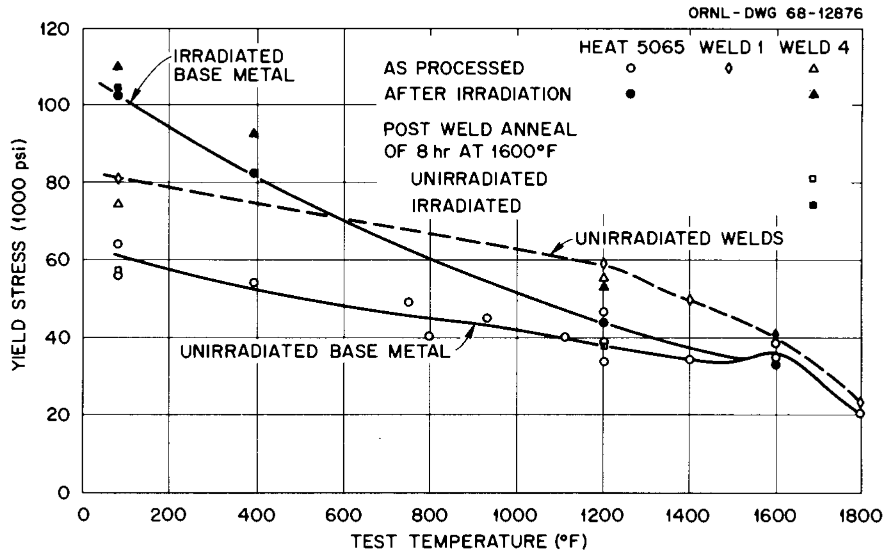


Fig. 3. Yield Strength of Hastelloy N Welds and Base Metal in Tensile Tests at a Strain Rate of 0.05 min^{-1}.

recovery. Thus, above 1200°F the yield strengths of base metal and welds annealed for 8 hr at 1600°F are equivalent and are unaffected by irradiation.

Figure 4 shows that the ultimate tensile strength is about the same for welded samples and for base metal. This property is not affected significantly by postweld heat treatment at 1600°F. Irradiation causes a slight (approx 10%) increase in the ultimate tensile strength at the lower test temperatures and a decrease at test temperatures of 1200°F and greater. This reduction at higher temperatures is due to the reduced ability of the material to deform plastically after irradiation (i.e., fracture occurs before the stress increases to the higher values noted for unirradiated materials).

Thus, the effects of welding on the strength parameters measured by standard tensile tests are relatively small and the most significant factor is the reduction in the fracture strain. The fracture strains observed for the welds involved in this study are summarized in Fig. 5.

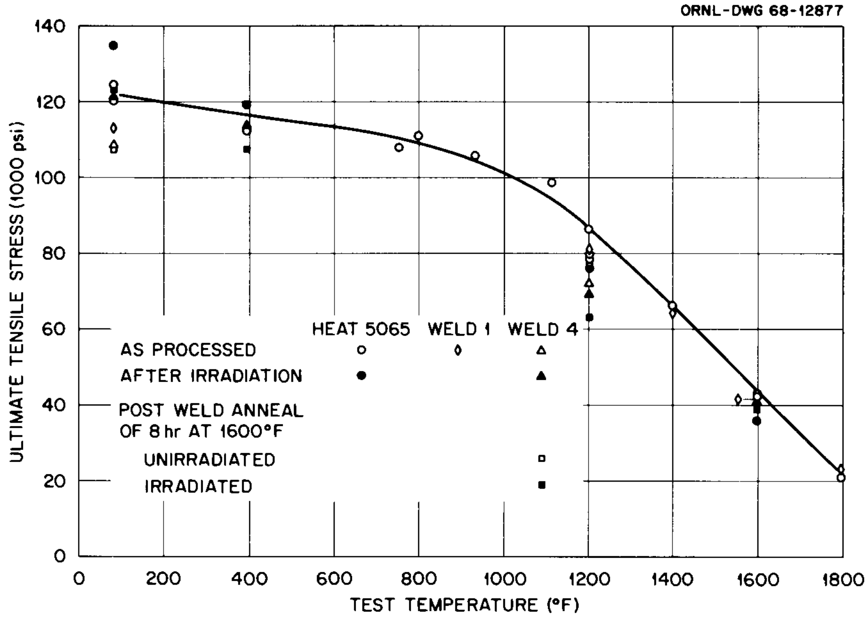


Fig. 4. Ultimate Tensile Strength of Hastelloy N Welds and Base Metal in Tensile Tests at a Strain Rate of 0.05 min^{-1}.

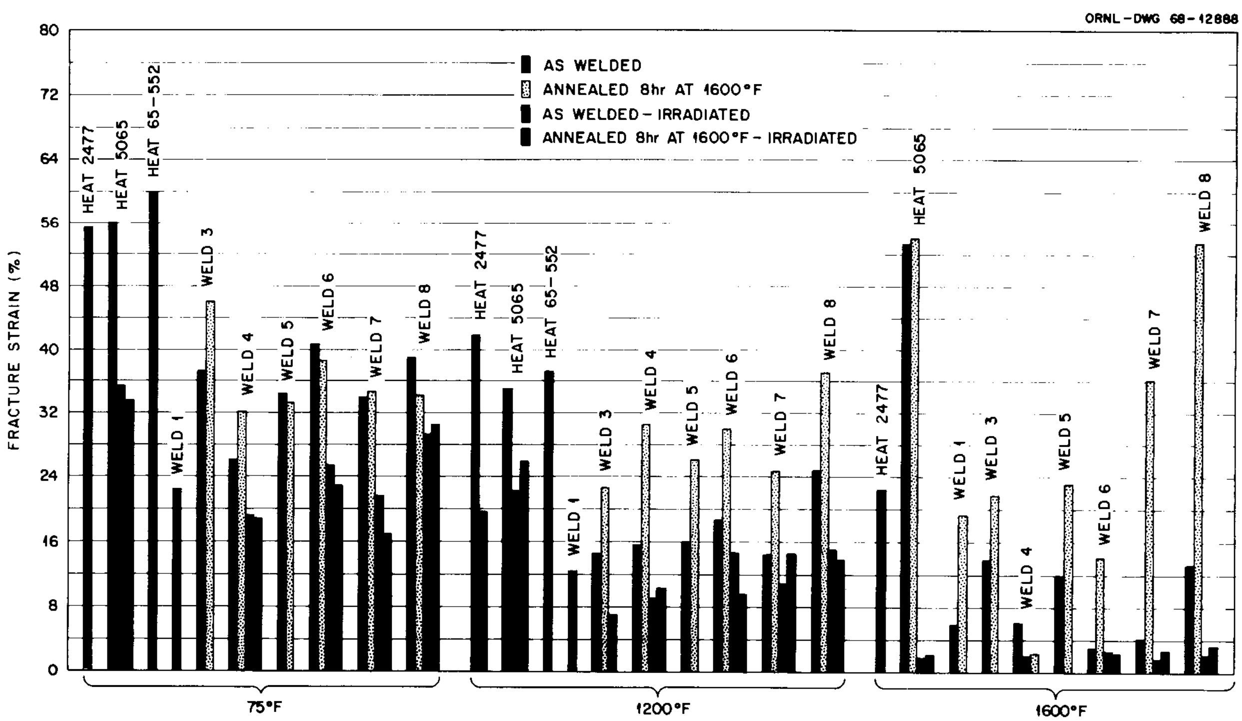


Fig. 5. Comparison of the Fracture Strains of Welds and Base Metal in Tensile Tests at a Strain Rate of 0.05 min^{-1}.

Welds 1, 4, and 5 involve air-melted materials with relatively high boron levels. Heat 2477, a vacuum-melted heat containing 8 ppm B, was used as the base and filler metal in weld 3. Weld 6 utilized filler metal from heat 65-552, a vacuum-melted heat containing 1 ppm B, and base metal from heat 5065, an air-melted heat. Welds 7 and 8 utilized air-melted base and filler materials. The filler rod for weld 7 was plasma-spray-coated with Al_2O_3 in an effort to reduce the interdendritic spacing in the cast weld metal and to provide additional sites for helium collection. However, most of the Al_2O_3 floated on the weld and had to be removed mechanically before making the next pass; the aluminum content of the deposited weld metal was only 0.06% compared with 0.84% for a typical cross section from the filler metal before deposition. Weld 8 involved a filler material coated with WC. The deposited weld metal contained 2% W and 0.1% C, so the WC coating dissolved in the melt altered the properties of the weld appreciably.

The fracture strains shown in Fig. 5 are compared for

1. base metal given a standard solution anneal of 1 hr at 2150°F,
2. base metal given the 1 hr at 2150°F treatment followed by an anneal of 8 hr at 1600°F,
3. transverse weld samples in the as-welded condition, and
4. transverse weld samples annealed for 8 hr at 1600°F.

These properties are considered for unirradiated and irradiated conditions where data are available.

At a test temperature of 75°F the following points are illustrated in Fig. 5.

1. The fracture strains of all welds are lower than those of the base metal.

2. Welds 1, 4, and 5 all involved air-melted alloys and exhibited the lowest fracture strains.

3. Welds 3 and 6 involved vacuum-melted weld metal and have the best properties.

4. Welds 7 and 8, which involved air-melted alloys with the filler metal modified with Al_2O_3 and WC, exhibited intermediate properties.

5. Postweld heat treatments generally improved the properties.

6. Irradiation decreased the fracture strain of all welds and base metal. The welds had lower postirradiation ductilities than the base metals, but the welds were affected less by irradiation.

7. The postirradiation ductilities of welds did not depend appreciably on whether they had received a postweld heat treatment prior to irradiation.

8. The postirradiation properties of all welds except No. 8 were equivalent.

Figure 5 also illustrates several important points at a tensile test temperature of 1200°F.

1. The fracture strains of all welds were lower than those of the base metal.

2. The as-welded ductilities of all welds except No. 8 were about equivalent, ranging from 12 to 18%. Weld 8 had 25% strain at fracture under these conditions.

3. A postweld heat treatment of 8 hr at 1600°F markedly improved the fracture strain, with values of 22 to 30% being observed. Weld 8 again was better.

4. After irradiation, all welds were less ductile than the base metal. There were some differences in the ductilities of the welds, but no consistent trends are apparent.

5. Postweld annealing for 8 hr at 1600°F had little effect on the postirradiation properties.

At 1600°F, Fig. 5 shows the following trends.

1. The fracture strains were drastically lower for welds than for base metal. Welds 3 and 8 had properties superior to those of the other welds. Postweld heat treating had a beneficial effect on the ductility.

2. Irradiation reduced the fracture strain to about 2% independent of whether the test sample was base metal or a transverse weld.

All of the postirradiation results presented thus far have been for materials irradiated at less than 300°F. We have irradiated several samples of these heats and others at 1200 to 1400°F and find that there are at least two significant differences in the results obtained. The displacement damage anneals at these irradiation temperatures and the tensile properties up to about 1000°F are the same for irradiated and unirradiated materials. At higher test temperatures the fracture strain is reduced even further if the irradiation temperature is in the range of 1200 to 1400°F. For example, heat 5065 was found to have a fracture strain of 11.3% (irradiated at 1200°F) at 1200°F compared with the value of 22.2% (irradiated at less than 300°F) shown in Fig. 5. Weld 7 was observed to have fracture strains at 1200°F of 7.2 and 10.8% after

irradiation at 1200°F and less than 300°F, respectively. Thus, the trends shown in Fig. 5 for samples irradiated at 300°F seem to hold for irradiation temperatures in the 1200 to 1400°F range, although the actual fracture strains may be lower for the higher irradiation temperature.

Creep-Rupture Properties*

The stress-rupture properties at 1200°F of several heats of base metal are shown in Fig. 6. The data are described reasonably well by a single line, although both air- and vacuum-melted materials are involved. The line for the base materials is shown in Fig. 7 where a comparison is made between the stress-rupture properties of the base metal and the various welds involved in this study. All of the welds had lower creep-rupture strength in the as-welded condition than the base metal. Weld 3, which involved only vacuum-melted materials, had significantly higher strength than the other welds. Welds 7 and 8, which involved Al_2O_3 and WC additions, also showed notably better performance than welds 1, 4, and 6. The rupture lives of the welds approached those of the base metal as the stress level was reduced; this is likely due to thermal recovery of the weld metal during the long time at 1200°F. The strength was improved greatly by a postweld anneal of 8 hr at 1600°F.

The fracture strains of the welds and several heats of base metal are compared in Fig. 8. The fracture strain of the base metal was about 25% for short rupture lives of a few hours and decreased to values of 10 to 15% when the rupture life was 1000 hr. There is no appreciable difference between air- and vacuum-melted alloys. The welds had fracture

---

*See Tables A-3 and A-4, Appendix, for the tabulated creep-rupture data.

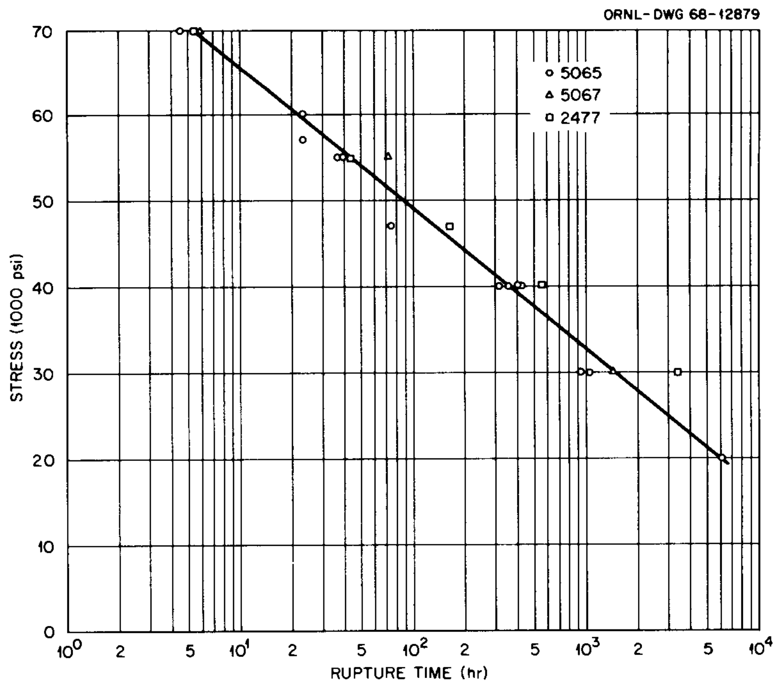


Fig. 6. Stress-Rupture Properties of Hastelloy N Base Metal at 1200°F.

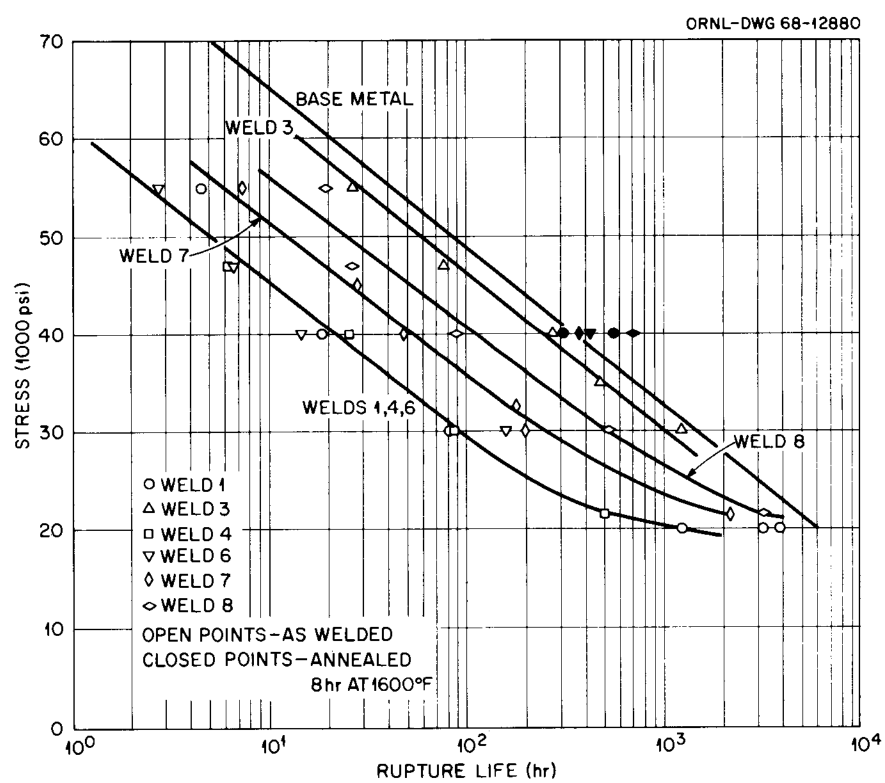


Fig. 7. Stress-Rupture Properties of Several Hastelloy N Welds at 1200°F.

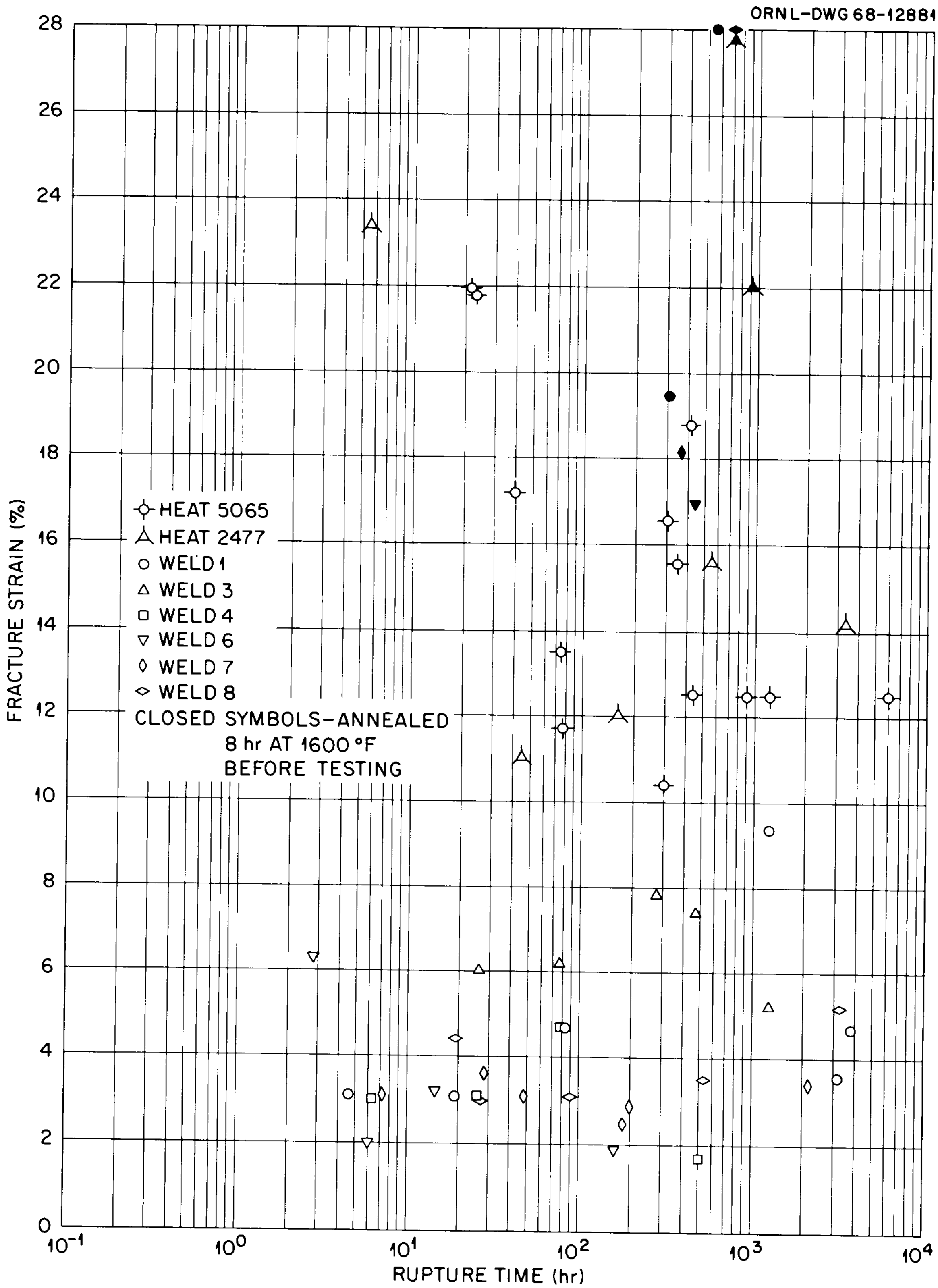


Fig. 8. Variation of Fracture Strain with Rupture Life for Welds and Base Metal Tested at 1200°F.

strains of only about 3% for rupture lives up to a few hundred hours. Samples having rupture lives over 1000 hr showed some improvement with strains in the range of 3 to 5%. Weld 3 had much superior fracture strains of about 6% for all test conditions. All welds and even the base metal exhibited much higher fracture strains after an anneal of 8 hr at 1600°F.

As shown in Fig. 9, irradiation produces some rather dramatic changes in the stress-rupture properties. The lines shown in Fig. 9 were obtained from Fig. 6 for the unirradiated base metal, Fig. 8 for the minimum strength of unirradiated welds, and from Ref. 6 for the strength of irradiated base metal. Several points are included for base metals irradiated under exactly the same conditions as the welds. The properties of heat 2477 (low boron, vacuum-melted) are superior to

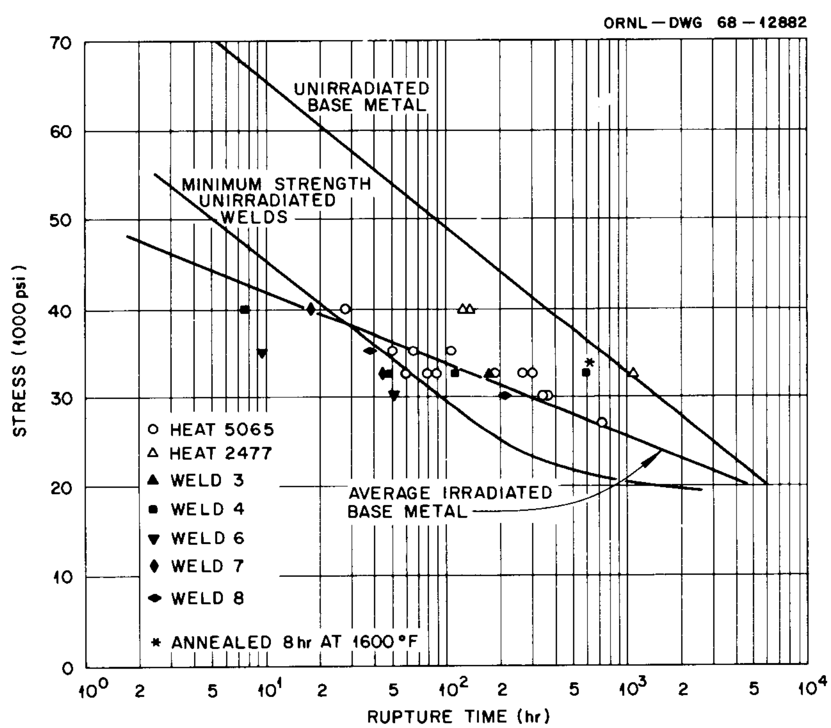


Fig. 9. Influence of Irradiation on the Stress-Rupture Properties of Hastelloy N Welds and Base Metal at 1200°F. All welds in the as-welded condition unless otherwise indicated.

those of heat 5065 for the low irradiation temperature of less than 300°F, but further work has shown that these materials have comparable properties when irradiated at 1200°F or higher.[6] The rupture lives of the welds seem quite comparable with those of the base metals and all welds seem to have about the same properties. A single specimen was given a postweld heat treatment of 8 hr at 1600°F and had superior postirradiation strength, but this observation should be repeated before concluding that this anneal is effective. Moteff and Smith[15] also found that the postirradiation creep properties of Hastelloy N base metal and welds were equivalent at 1200°F. They also found that a postweld anneal of 4 hr at 1600°F did not improve the postirradiation properties of welds.

The fracture strain is shown in Fig. 10 as a function of rupture life. The fracture strain of the base metal had a low value of about 0.5% for rupture lives of only a few hours and increased to about 3% for rupture lives of 1000 hr. The welds fall into this same pattern, but the strains were consistently slightly lower. Again the single test on a sample having a postweld anneal exhibited better properties.

We also compared the minimum creep rates for these same materials to obtain a measure of strength. The creep properties of several heats of base metal in the irradiated and unirradiated conditions are shown in Fig. 11. Most of the data are contained within the band in Fig. 11, with heat 5065 favoring the right side and heat 2477 the left side. This same band is transferred to Fig. 12, where the data for the welds are shown for comparison. The data do not deviate significantly from the band obtained for the base metal. Thus, the creep rate of Hastelloy N is not affected appreciably by welding or by irradiation.

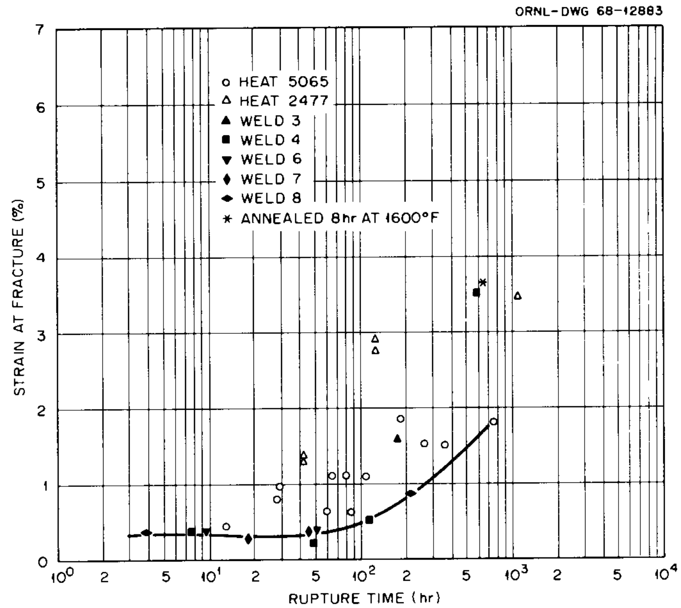


Fig. 10. Influence of Irradiation on the Fracture Strain of Hastelloy N Welds and Base Metal in Creep at 1200°F. All welds in the as-welded condition unless otherwise indicated.

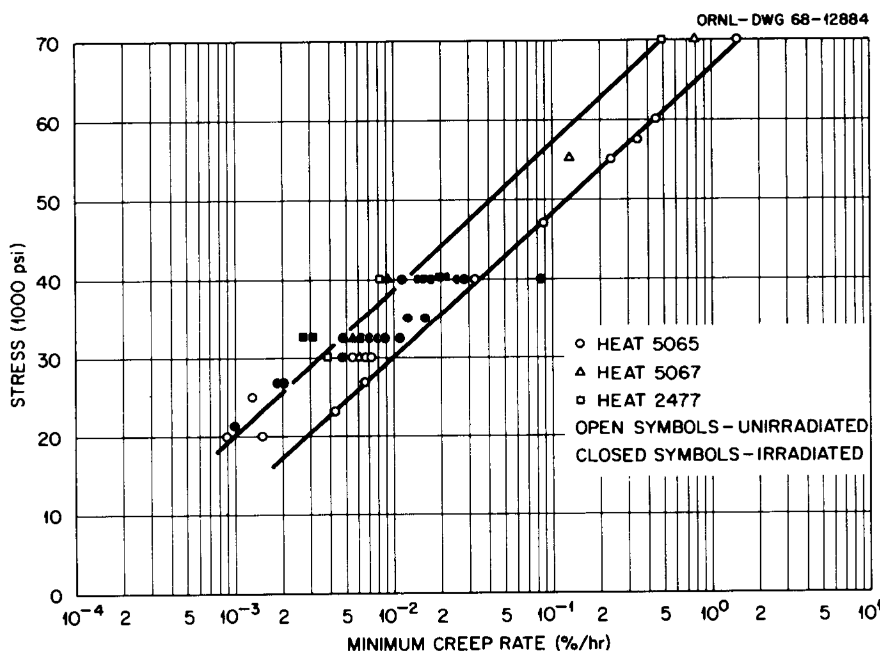


Fig. 11. Influence of Irradiation on the Creep Rate of Hastelloy N Base Metal at 1200°F.

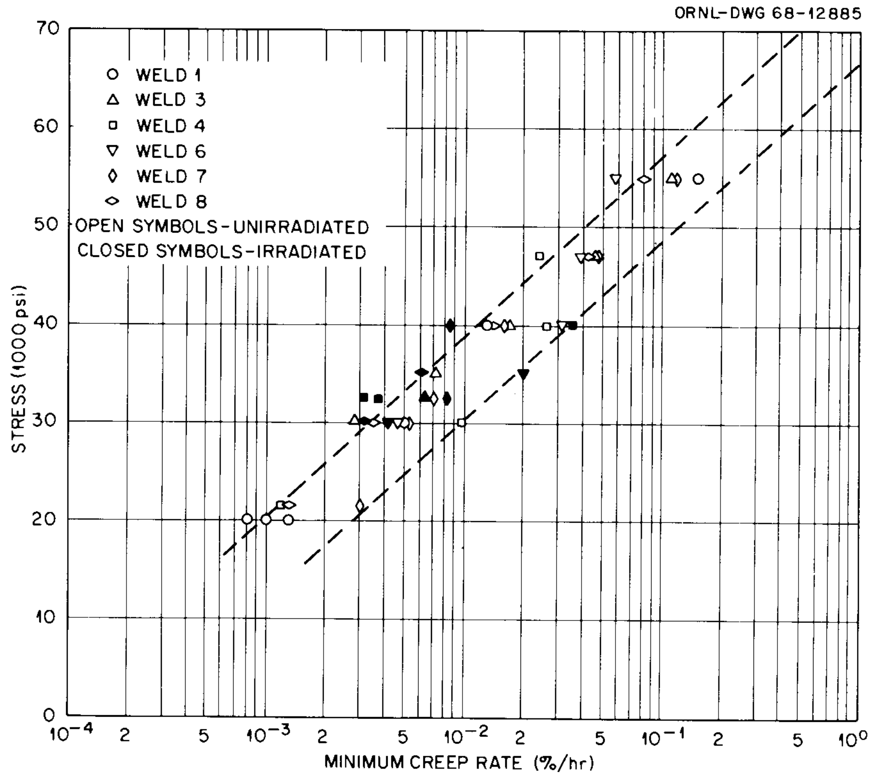


Fig. 12. Influence of Irradiation on the Creep Rate of Hastelloy N Welds at 1200°F.

Postweld Heat Treatments*

We have mentioned previously the improvement of the properties of welds by a postweld heat treatment of 8 hr at 1600°F. We chose this treatment on the basis of our work with weld 1 in which we tested samples at 40,000 psi and 1200°F after annealing samples for various periods of time at temperatures of 1200, 1400, and 1600°F. The influence of these treatments on the rupture life is shown in Fig. 13. The as-welded sample had a rupture life of only 19 hr compared with a rupture life of about 350 hr for the base metal (Fig. 6). Long annealing times at 1200°F were required to produce much improvement, and the rupture life was only 60 hr

*See Table A-5, Appendix, for the tabulated results for creep-rupture tests.

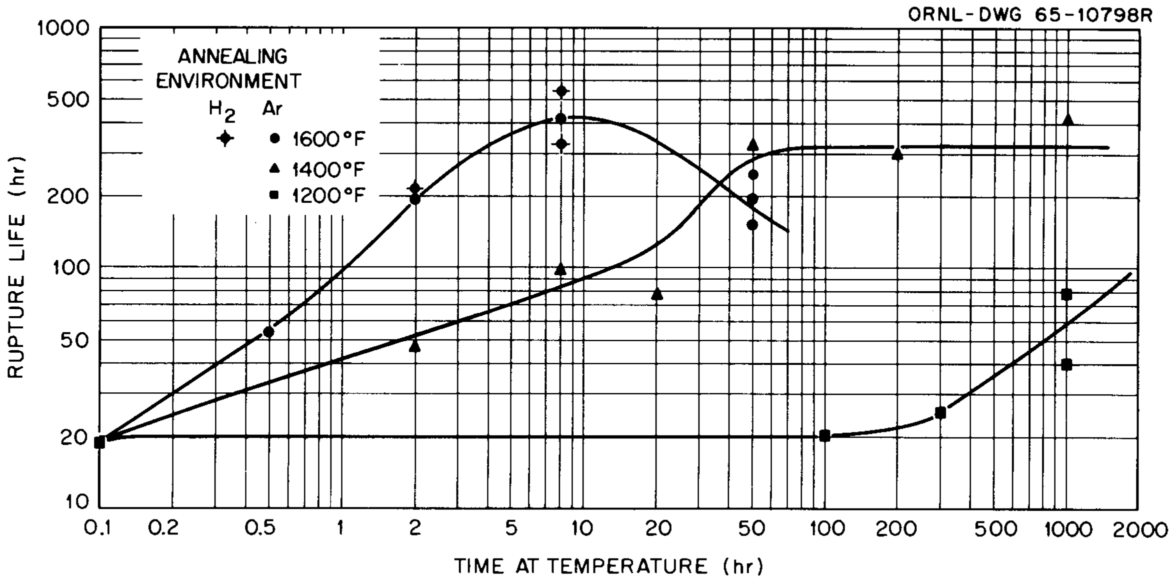


Fig. 13. Influence of Postweld Heat Treatments on the Rupture Life of Weld 1 Tested at 1200°F and 40,000 psi.

after a 1000-hr postweld heat treatment at 1200°F. Annealing at 1400°F brought about more rapid recovery with the rupture life of the weld being comparable with that of the base metal after a 50-hr anneal. Only about 8 hr at 1600°F was required to restore the rupture life. We have no explanation for the apparent reduction in the rupture life due to a postweld anneal at 1600°F of 50 hr compared with anneals of 8- and 100-hr duration.

The change in the fracture strain with postweld heat treatment is shown in Fig. 14. The fracture strain is about 12% for the base metal (Fig. 8) and the welds can be annealed to obtain equivalent or even better fracture strains. There is an apparent inversion in the recovery curve at 1600°F; this corresponds to the conditions producing short rupture lives in Fig. 13.

The reduction in area for these same samples is shown in Fig. 15. Values for the welds as high as the 15% obtained for base metal can be

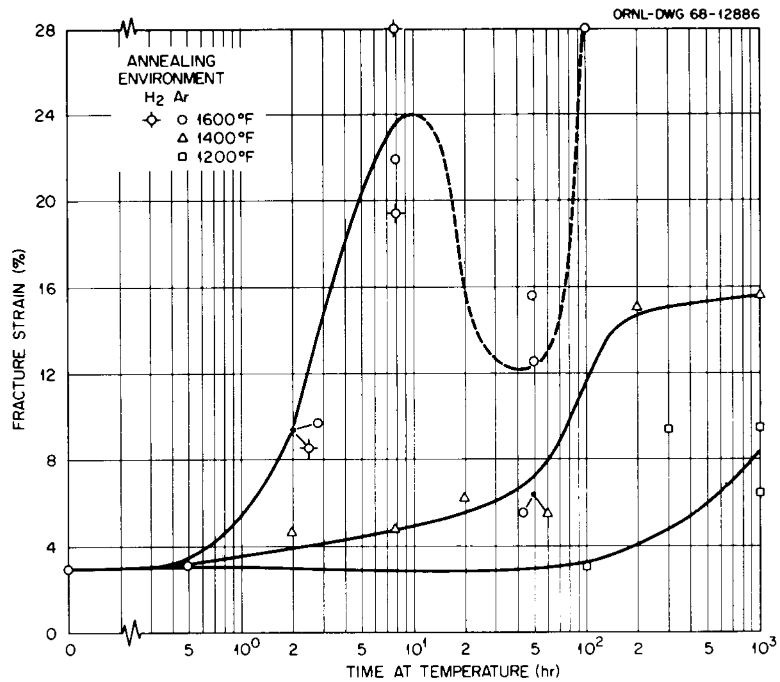


Fig. 14. Influence of Postweld Heat Treatments on the Fracture Strain of Weld 1 Tested at 1200°F and 40,000 psi.

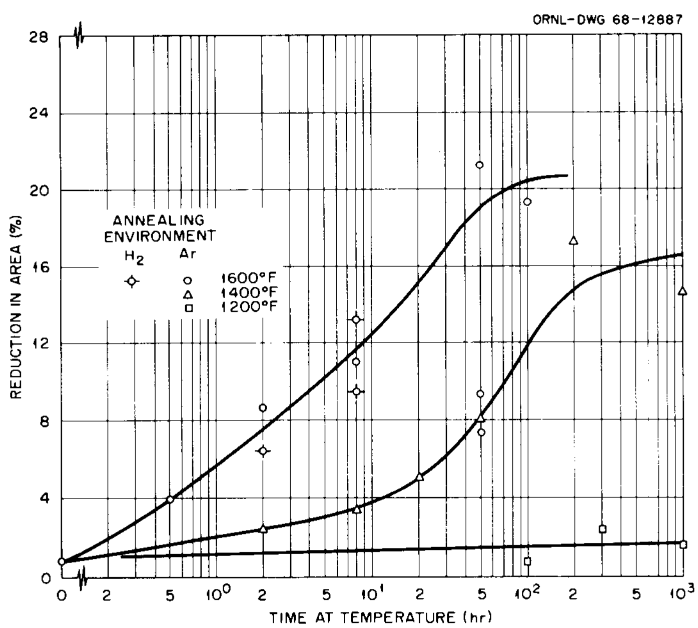


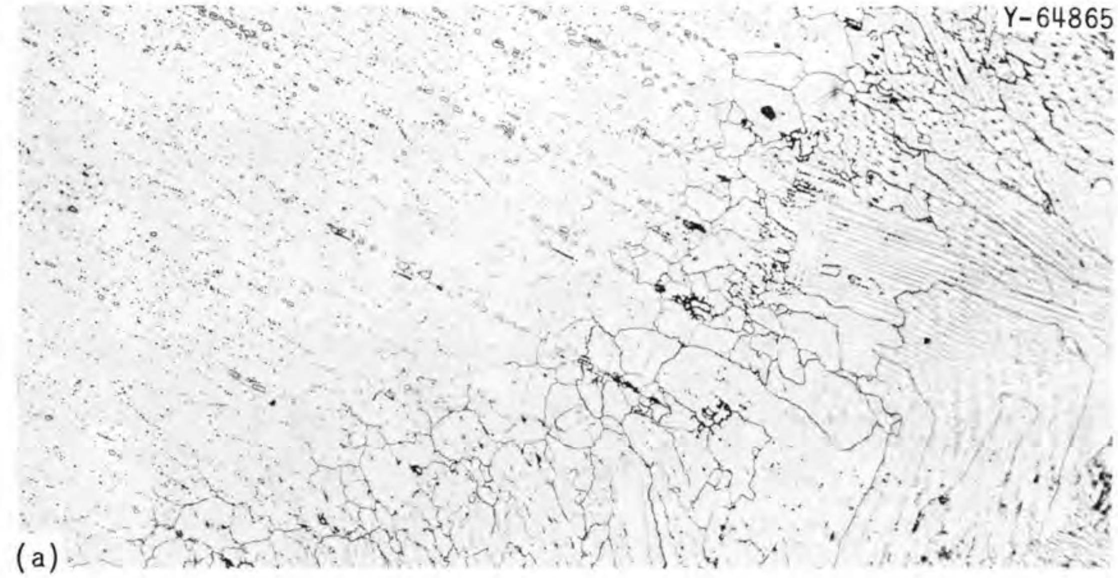
Fig. 15. Influence of Postweld Heat Treatments on the Reduction in Area of Weld 1 Tested at 1200°F and 650°C.

obtained by postweld anneals at 1400 and 1600°F. The unusual behavior due to annealing for 50 hr at 1600°F again appears for two of the samples. The reduction in area is not affected appreciably by annealing at 1200°F, indicating that the improved elongation at fracture (Fig. 14) is due primarily to the formation of cracks during testing.

We examined the influence of whether the postweld annealing environment was argon or hydrogen at 1600°F, since Gilliland and Venard[3] had reported better properties for samples annealed in hydrogen than for those annealed in argon. Comparative points are shown in Figs. 13, 14, and 15 for these annealing environments, and no appreciable differences are apparent.

Metallography

Many of the samples were examined metallographically and a few typical microstructures will be presented. Figure 16 shows the fusion zone of weld 4. The structure of the base metal is characterized by stringers of M_6C-type precipitates and a relatively fine grain size. The weld metal contains a fine cellular structure within coarse grains. The fusion line is quite sharp and the transition of some of the large M_6C precipitates to a lamellar phase is apparent. The exact identification of the lamellar phase has eluded us for some time, but our evidence indicates that it is a Mo-Ni-Si compound.[16] Microprobe studies have shown that the precipitates are enriched in silicon and that the dark etching portions of the dendritic cast structure are also higher in silicon.



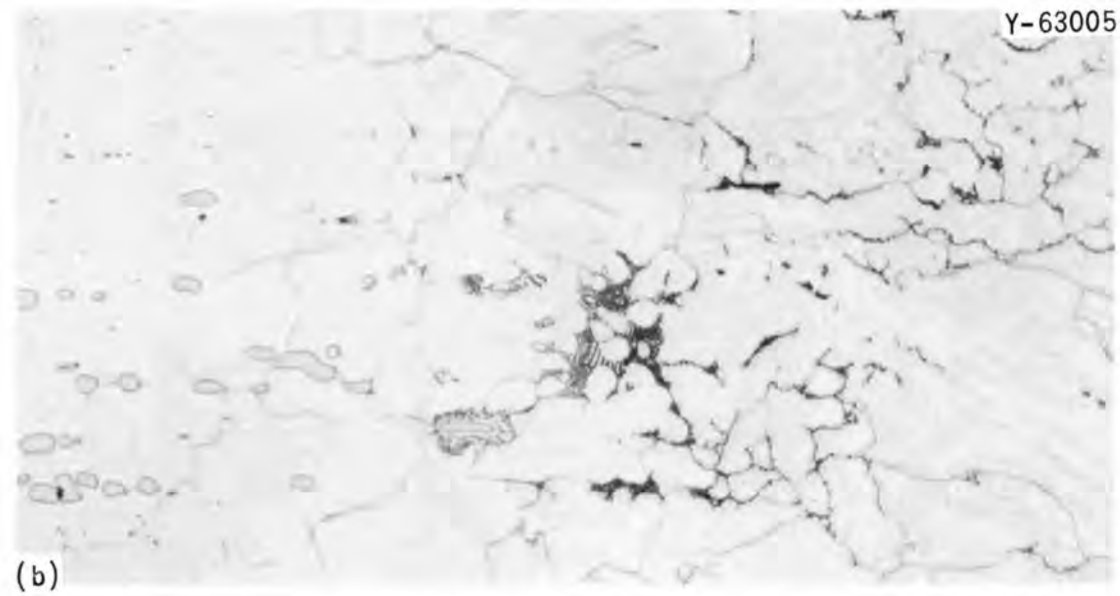


Fig. 16. Photomicrographs of the Fusion Zone of Weld 4. (a) Base metal on left and weld metal on right. 100×. (b) Transformation of stringer precipitates to lamellar product at the fusion line. 500×. Etchant: glyceria regia.

This same general description applies to the fusion zones of the other welds except 3 and 8. The base metals used in weld 3 were vacuum-melted and the stringers are minimal. The bulk silicon content was low (0.015%) and the precipitates were primarily of the Mo_2C type and were found to dissolve easily. Thus, not much of the lamellar transformation

product was present at the fusion line. Weld 8 involved the filler material that was coated with WC, and the structure of the deposited weld metal is much finer (Fig. 17).

A typical microstructure of a creep-rupture fracture in Hastelloy N base metal is shown in Fig. 18. This sample was tested at 40,000 psi and 1200°F and failed after 312 hr with 16.6% strain. There is considerable grain boundary cracking and the fracture is predominantly intergranular.

Typical microstructures of a tested sample from weld 1 are shown in Fig. 19. This sample was tested at 40,000 psi and 1200°F in the as-welded condition; failure occurred in 18.7 hr with 3.1% strain and only 0.77% reduction in area. The fracture occurred in the weld metal and followed the grain boundaries; however, little grain boundary cracking adjacent to the fracture is evident.

Postweld heat treatments generally improved the tensile and creep properties. Figure 20 shows the fracture of a sample from weld 1 that was annealed for 2 hr at 1600°F and tested at 40,000 psi and 1200°F. This sample failed in 200 hr with 9.4% strain and 6.4% reduction in area. The failure still occurred in the weld metal; however, intergranular cracks did form throughout the weld zone, indicating that deformation occurred generally throughout this zone.

SUMMARY

This study has shown that the mechanical properties of welds in Hastelloy N are generally inferior to those of the base metal. The strain at fracture is the property affected most severely, although appreciable reductions in the high-temperature creep-rupture strength

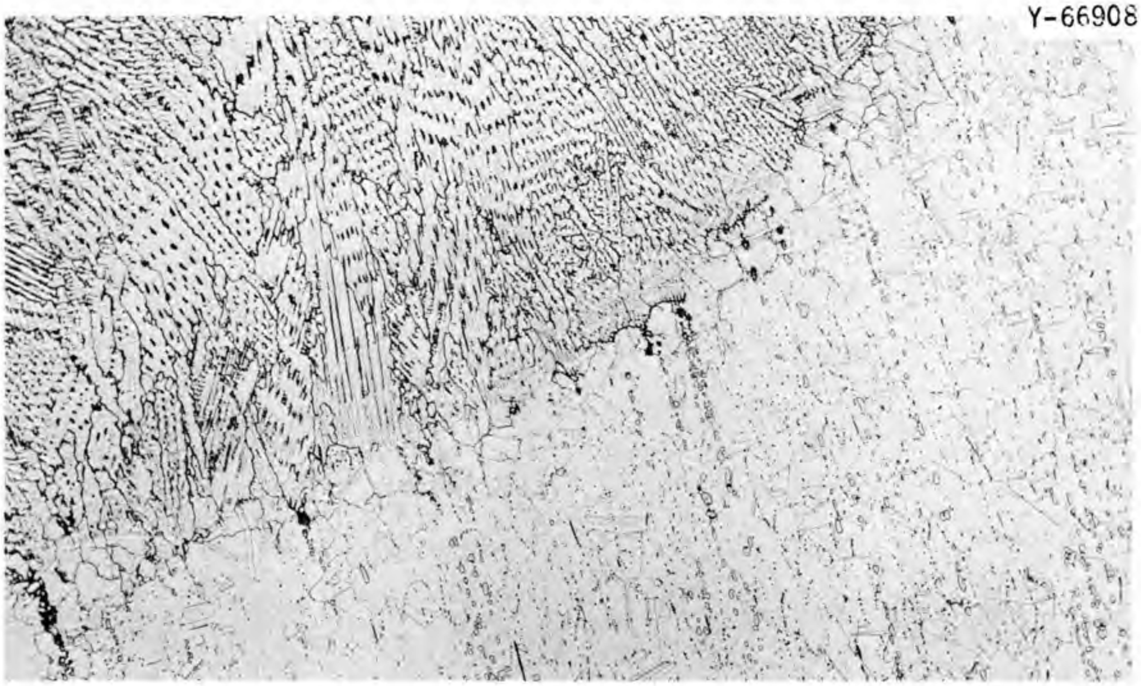


Fig. 17. Photomicrograph of the Fusion Line of Weld 8. Weld metal in the upper left and base metal in the lower right. 100×. Etchant: glyceria regia.

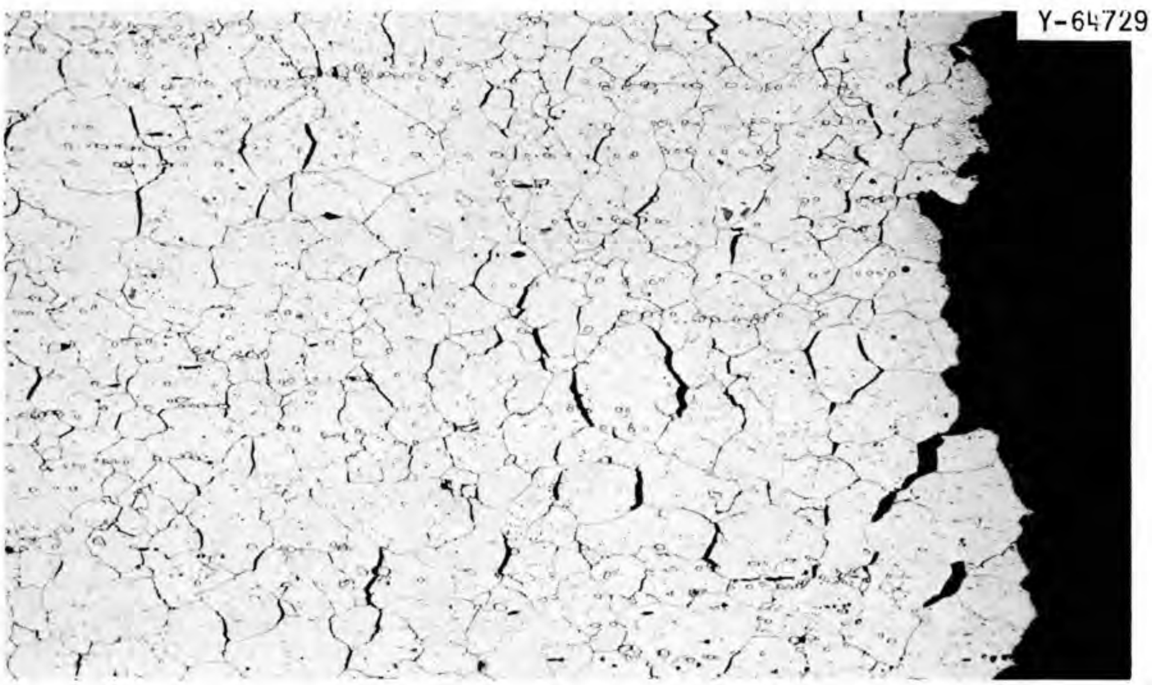


Fig. 18. Photomicrograph of the Fracture of a Hastelloy N (Heat 5065) Sample Tested in Creep at 40,000 psi and 1200°F Following a Pretest Anneal of 1 hr at 2150°F. Failed in 312 hr with 16.6% strain. 100×. Etchant: glyceria regia.

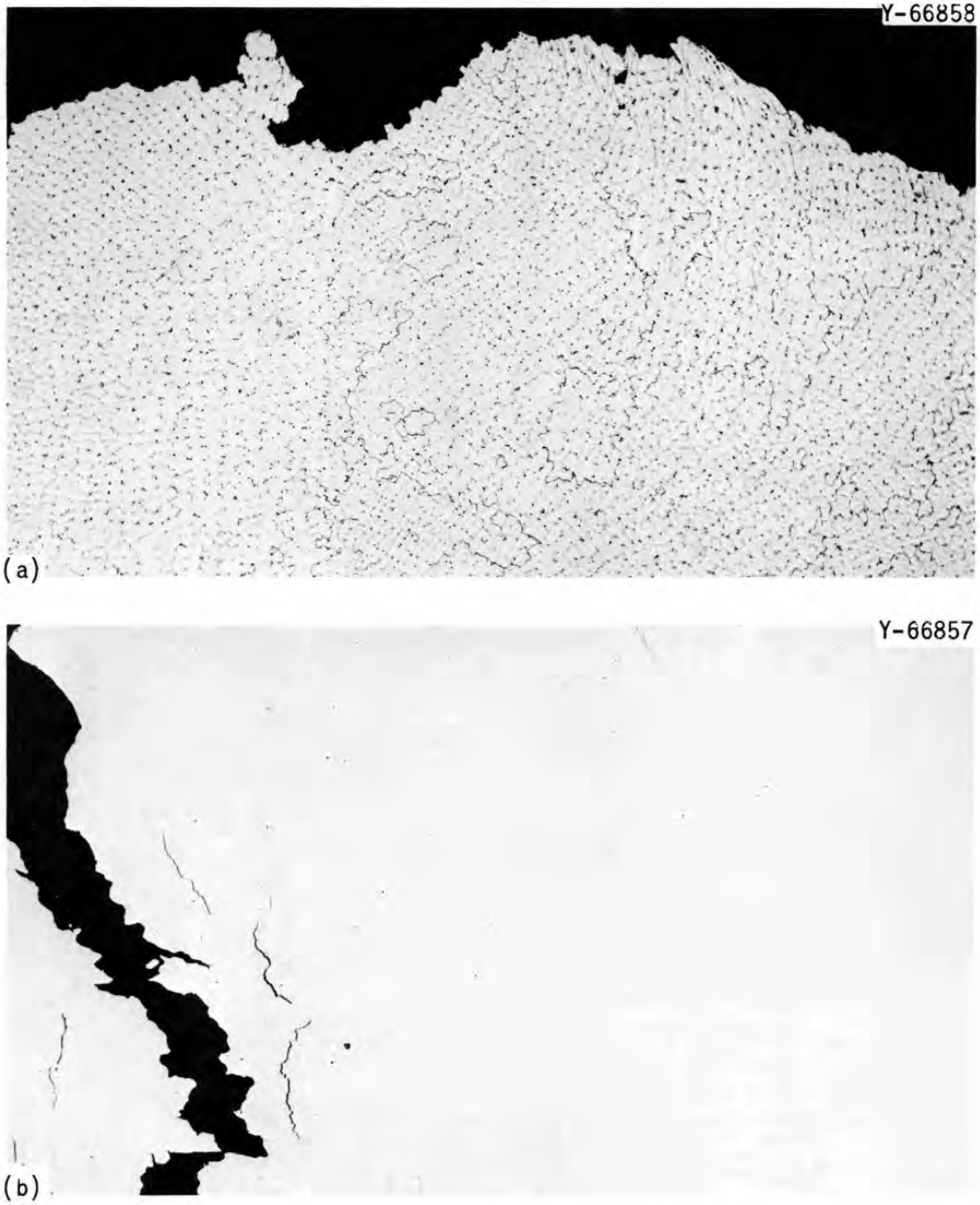


Fig. 19. Photomicrographs of Sample from Weld 1 Tested at 40,000 psi and 1200°F in the As-Welded Condition. Failed in 18.7 hr with 3.1% strain. 100×. (a) Fracture in the weld metal. Etchant: glyceria regia. (b) An unetched photomicrograph of a cracked area away from the fracture.

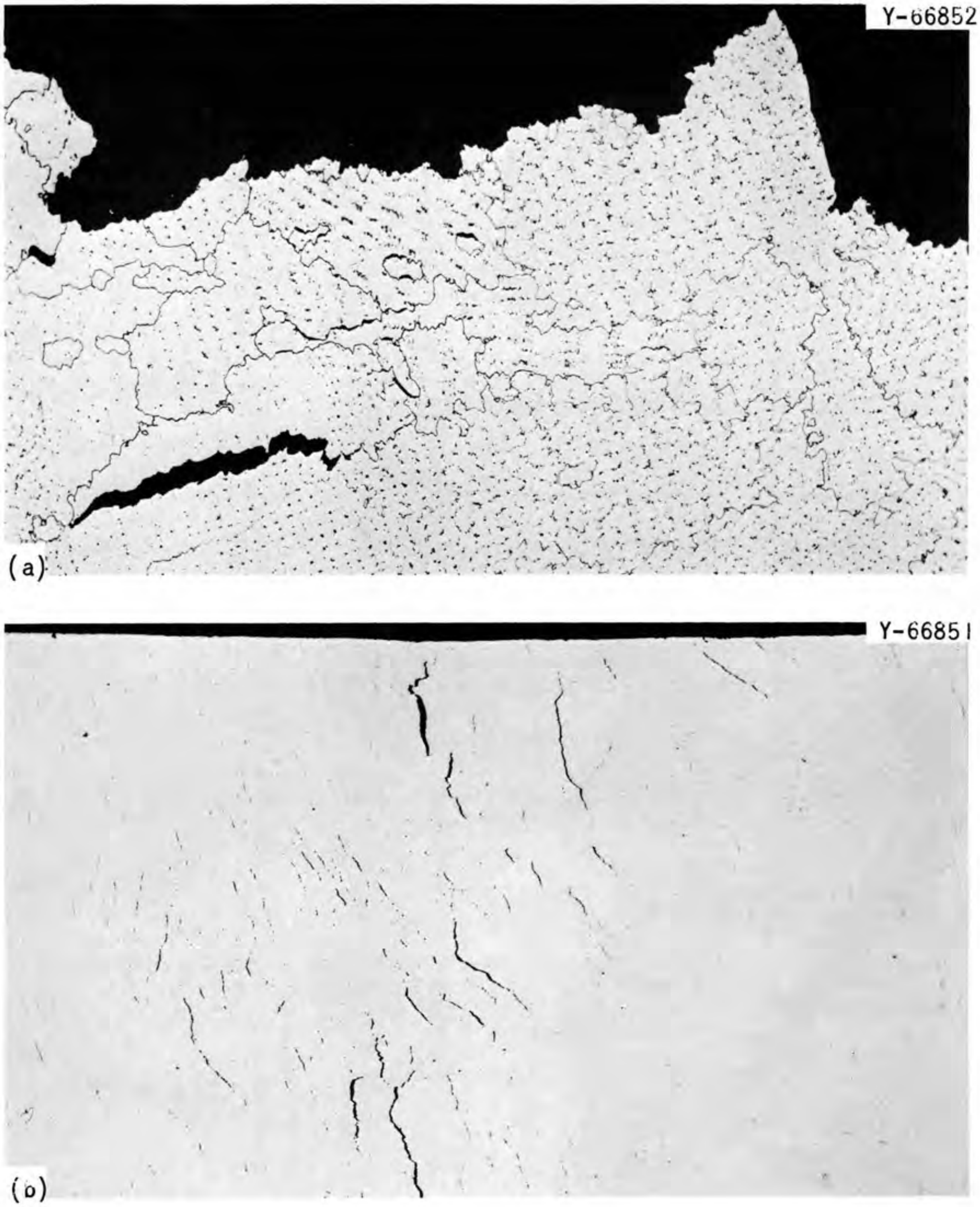


Fig. 20. Photomicrographs of Tested Creep Sample from Weld 1. Sample given a postweld anneal of 2 hr at 1600°F in argon and tested at 40,000 psi and 1200°F. Failed in 200 hr with 9.4% strain. 100×. (a) Fracture. Etchant: glyceria regia. (b) An unetched photomicrograph of deformed area in the weld metal.

also occurred. The properties of welds can be improved by suitable postweld heat treatments to almost equal those of the base metal.

Several welds were studied which involved air- and vacuum-melted materials. The weld involving vacuum-melted base and filler metal (weld 3) had superior creep-rupture properties in the as-welded condition. Weld 8 involved air-melted materials and the filler metal was coated with WC. The properties of this weld were generally superior to those of the other welds, but the fracture strain under creep conditions was not improved.

Neutron irradiation caused a decrease in the fracture strain of both welds and base metal. The postirradiation fracture strains in tensile tests of the welds and the base metal became closer as the test temperature was increased; they were identical at 1600°F. Although postweld heat treatments improved the fracture strain in preirradiation tensile tests, the postirradiation fracture strains of welds seemed independent of postweld anneal. The postirradiation creep-rupture properties at 1200°F were about the same for welds and base metal, even though the preirradiation properties differed greatly. Although some of the irradiated welds showed improved performance in tensile tests, they all had about the same creep-rupture properties.

ACKNOWLEDGMENTS

The welds were made under the supervision of T. R. Housley and the mechanical property tests were run by B. C. Williams, E. Bolling, N. O. Pleasant, J. T. Feltner, and V. G. Lane. H. R. Tinch was responsible for the metallographic work. We are also indebted to

G. M. Slaughter and J. R. Weir for their interest in this work and for their careful review of the manuscript. The drawings were prepared by the Graphic Arts Department and the manuscript was prepared by the Metals and Ceramics Division Reports Office.

REFERENCES

1. W. D. Manly *et al.*, "Metallurgical Problems in Molten Fluoride Systems," Progr. Nucl. Energy Ser. IV **2**, 164–179 (1960).

2. H. G. MacPherson, "Molten Salt Reactor Shows Most Promise to Conserve Nuclear Fuels, Part 2," Power Eng. **71**(2), 56–58 (1967).

3. R. G. Gilliland and J. T. Venard, "Elevated Temperature Mechanical Properties of Welds in Ni–Mo–Cr–Fe Alloy," Welding J. (N.Y.) **45**(3), 103-s–110-s (1966).

4. R. G. Gilliland and G. M. Slaughter, private communication.

5. W. R. Martin and J. R. Weir, "Postirradiation Creep and Stress Rupture of Hastelloy N," Nucl. Appl. **3**, 167 (1967).

6. H. E. McCoy, "Variation of the Mechanical Properties of Irradiated Hastelloy N with Strain Rate," submitted to Journal of Nuclear Materials.

7. D. R. Harries, "Neutron Irradiation Embrittlement of Austenitic Stainless Steels and Nickel Base Alloys," J. Brit. Nucl. Energy Soc. **5**, 74 (1966).

8. G. H. Broomfield, D. R. Harries, and A. C. Roberts, "Neutron Irradiation Effects in Austenitic Stainless Steels and a Nimonic Alloy," J. Iron Steel Inst. (London) **203**, 502 (1965).

9. F. C. Robertshaw et al., "Neutron Irradiation Effects in A-286 Hastelloy Y and René 41 Alloys," Spec. Tech. Publ. No. 341, p. 372, American Society for Testing and Materials, Philadelphia, Pa., 1963.

10. N. E. Hinkle, "Effect of Neutron Bombardment on Stress-Rupture Properties of Some Structural Alloys," Spec. Tech. Publ. No. 341, p. 344, American Society for Testing and Materials, Philadelphia, Pa., 1963.

11. P.C.L. Pfeil and D. R. Harries, "Effects of Irradiation in Austenitic Steels and Other High-Temperature Alloys," p. 202 in Flow and Fracture of Metals and Alloys in Nuclear Environments Spec. Tech. Publ. No. 380, American Society for Testing and Materials, Philadelphia, Pa., 1965.

12. J. T. Venard and J. R. Weir, "In-Reactor Stress-Rupture Properties of a 20 Cr–25 Ni Columbium-Stabilized Stainless Steel," p. 269 in Flow and Fracture of Metals and Alloys in Nuclear Environments Spec. Tech. Publ. No. 380, American Society for Testing and Materials, Philadelphia, Pa., 1965.

13. P.C.L. Pfeil, P. J. Barton, and D. R. Arkell, "Effects of Irradiation on the Elevated Temperature Mechanical Properties of Austenitic Steels," Trans. Am. Nucl. Soc. 8, 120 (1965).

14. P.R.B. Higgins and A. C. Roberts, "Reduction in Ductility of Austenitic Stainless Steel after Irradiation," Nature 206, 1249 (1965).

15. J. Moteff and J. P. Smith, Sixth Annual Report of High Temperature Materials Program, Part A, March 31, 1967, GEMP-475A, p. 185 (1967).

16. R. E. Gehlbach and H. E. McCoy, "Phase Instability in Hastelloy N," paper presented at the International Symposium on Structural Stability in Superalloys, Seven Springs, Pa., Sept. 4–6, 1968; to be published in the proceedings.

APPENDIX

Table A-1. Tensile Properties of Welds in the Unirradiated Condition

Specimen Number	Weld Number	Postweld Anneal	Strain Rate (min^{-1})	Test Temperature (°F)	Stress, psi		Elongation, %		Reduction in Area (%)
					Yield	Ultimate	Uniform	Total	
13	1	None	0.05	75	81,100	113,100	22.0	22.3	37.7
14	1	None	0.05	390	59,100	81,200	11.3	12.3	24.4
15	1	None	0.05	1200	50,000	64,800	6.0	7.8	13.1
16	1	None	0.05	1200	40,800	42,000	3.0	5.8	6.9
17	1	None	0.05	1600	23,200	23,400	1.8	21.8	25.5
18	1	a	0.05	1600	41,600	42,100	3.9	19.2	18.6
3000	1	None	0.05	75	76,300	111,900	27.1	28.3	29.5
3001	1	None	0.05	1200	59,300	79,800	10.2	11.7	26.5
3005	3	None	0.05	75	71,700	122,800	35.2	37.2	42.4
3006	3	None	0.05	1200	53,200	77,300	12.6	14.4	23.6
10285	3	None	0.05	1600	41,300	44,900	4.4	13.8	14.8
10287	3	a	0.05	75	61,000	123,700	44.0	46.1	41.02
10286	3	a	0.05	1200	43,700	82,400	21.4	22.5	26.93
10300	3	a	0.05	1600	40,900	44,200	5.3	21.7	18.15
3007	4	None	0.05	75	74,300	108,900	24.9	25.9	47.8
3008	4	None	0.05	1200	55,300	72,300	13.5	15.3	31.7
10288	4	None	0.05	1600	40,900	44,700	3.1	5.9	7.66
10289	4	a	0.05	75	57,700	107,500	31.4	31.9	38.4
10290	4	a	0.05	1200	39,300	76,900	29.6	30.4	35.4
	5	None	0.05	75	73,300	119,800	33.0	34.2	45.1
	5	None	0.05	1200	51,100	70,100	14.2	15.9	42.8
10299	5	None	0.05	1600	43,300	45,400	1.7	11.7	18.0
10296	5	a	0.05	75	63,800	119,500	32.7	33.1	41.1
10297	5	a	0.05	1200	49,200	91,300	25.6	25.9	38.8
10295	5	a	0.05	1600	39,000	39,400	1.4	36.1	62.6
1144	6	None	0.05	75	68,200	120,900	38.7	40.7	38.3
1145	6	None	0.05	1200	49,800	81,300	16.6	18.6	20.9
1147	6	None	0.05	1600	39,000	44,400	2.2	2.9	3.04
1140	6	a	0.05	75	63,100	119,100	37.0	38.5	47.3
1141	6	a	0.05	1200	44,400	85,400	29.0	29.9	36.8
1142	6	a	0.05	1600	39,900	44,500	4.5	13.9	16.7
1164	7	None	0.05	75	71,900	125,700	32.3	33.8	45.0
1165	7	None	0.05	1200	52,000	79,100	13.7	14.4	11.9
777	7	None	0.002		46,900	65,700	7.3	8.6	18.7
1166	7	None	0.05	1600	41,500	45,100	2.0	3.9	7.8
10284	7	a	0.05	75	64,100	119,400	33.8	34.5	34.9
10283	7	a	0.05	1200	45,100	86,100	24.0	24.6	37.6
10295	7	a	0.05	1600	39,000	39,400	1.4	36.1	62.6
3011	8	None	0.05	75	72,300	124,900	35.8	38.8	49.4
	8	None	0.05	1200	54,000	91,300	22.9	24.9	29.1
778	8	None	0.002	1200	39,000	66,600	11.8	16.7	28.2
10280	8	None	0.05	1600	38,400	44,600	4.5	13.0	8.1
10282	8	a	0.05	75	58,900	120,100	33.9	34.0	28.4
10281	8	a	0.05	1200	41,700	91,200	35.7	37.0	37.7
10294	8	a	0.05	1600	39,800	43,800	4.3	53.5	64.4

[a]8 hr at 1600°F.

Table A-2. Tensile Properties of Welds in the Irradiated Condition

Specimen Number	Weld Number	Postweld Anneal	Experiment Number	Strain Rate (min^{-1})	Test Temperature (°F)	Stress, psi		Elongation, %		Reduction in Area (%)
						Yield	Ultimate	Uniform	Total	
469	3	None	ETR-41-30	0.05	1200	57,600	70,800	5.9	6.8	18.6
470	3	None	ETR-41-30	0.002	1200	56,400	62,200	3.0	3.6	12.6
699	4	None	ORR-149	0.05	75	109,900	120,500	16.8	19.2	48.7
700	4	None	ORR-149	0.05	392	92,900	113,300	18.9	19.1	25.3
701	4	None	ORR-149	0.05	1200	53,300	69,300	8.6	8.7	12.0
703	4	None	ORR-149	0.002	1200	51,900	63,400	4.7	5.2	12.8
702	4	None	ORR-149	0.05	1600	41,200	41,600	1.4	2.0	1.3
704	4	None	ORR-149	0.002	1600	26,000	26,000	0.9	2.0	9.2
708	4	a	ORR-149	0.05	75	104,500	122,800	17.2	18.7	14.5
709	4	a	ORR-149	0.05	392	82,700	107,600	21.0	22.2	27.3
710	4	a	ORR-149	0.05	1200	42,200	63,300	9.8	10.3	21.9
712	4	a	ORR-149	0.002	1200	45,400	58,300	5.6	6.6	28.1
711	4	a	ORR-149	0.05	1600	38,700	39,600	1.4	2.1	1.8
713	4	a	ORR-149	0.002	1600	25,300	25,300	1.0	2.0	5.9
457	4	None	ETR-41-30	0.05	1200	57,200	69,800	5.9	7.3	20.1
458	4	None	ETR-41-30	0.002	1200	57,900	64,900	2.5	3.3	11.3
461	4	a	ETR-41-30	0.05	1200	44,600	70,400	16.2	16.6	32.6
462	4	a	ETR-41-30	0.002	1200	43,200	59,700	8.9	11.0	15.6
714	6	None	ORR-149	0.05	75	111,000	130,700	22.6	25.4	47.0
715	6	None	ORR-149	0.05	392	88,800	112,100	20.4	22.4	27.4
716	6	None	ORR-149	0.05	1200	45,100	72,000	14.2	14.6	13.2
718	6	None	ORR-149	0.002	1200	51,700	63,300	6.1	7.0	30.4
717	6	None	ORR-149	0.05	1600	38,900	40,500	1.4	2.5	3.2
719	6	None	ORR-149	0.002	1600	24,900	24,900	0.8	1.4	6.0
723	6	a	ORR-149	0.05	75	106,500	127,100	21.1	22.7	33.4
724	6	a	ORR-149	0.05	392	83,900	111,200	25.0	25.9	34.9
725	6	a	ORR-149	0.05	1200	44,700	60,300	8.8	9.5	16.1

Table A-2. (continued)

Specimen Number	Weld Number	Postweld Anneal	Experiment Number	Strain Rate (min^{-1})	Test Temperature (°F)	Stress, psi		Elongation, %		Reduction in Area (%)
						Yield	Ultimate	Uniform	Total	
727	6	a	ORR-149	0.002	1200	43,500	53,600	4.6	5.4	9.2
726	6	a	ORR-149	0.05	1600	36,400	40,100	2.0	5.3	11.9
728	6	a	ORR-149	0.002	1600	18,600	18,900	0.9	1.1	5.1
529	7	None	ORR-149	0.05	75	112,400	135,100	18.4	21.4	25.7
530	7	None	ORR-149	0.05	392	89,700	114,000	18.8	20.1	34.0
531	7	None	ORR-149	0.05	1200	45,900	75,000	10.6	10.8	5.6
533	7	None	ORR-149	0.002	1200	53,400	65,200	5.0	6.4	5.6
532	7	None	ORR-149	0.05	1600	40,300	41,300	1.5	1.5	5.6
534	7	None	ORR-149	0.002	1600	25,100	25,100	1.0	1.6	5.9
538	7	a	ORR-149	0.05	75	107,700	122,900	14.6	16.7	54.2
539	7	a	ORR-149	0.05	392	88,100	110,500	19.1	20.3	31.8
540	7	a	ORR-149	0.05	1200	44,300	66,400	10.2	15.2	7.4
542	7	a	ORR-149	0.002	1200	45,100	56,600	4.2	4.6	10.4
541	7	a	ORR-149	0.05	1600	39,500	40,200	1.3	2.4	10.5
543	7	a	ORR-149	0.002	1600	24,800	24,800	1.0	1.4	2.4
1162	7	None	ETR-41-31	0.05	1200	50,000	65,600	7.2	7.2	
1163	7	None	ETR-41-31	0.002	1200	48,300	56,300	3.4	3.5	
729	8	None	ORR-149	0.05	75	110,300	137,100	25.0	29.0	48.91
730	8	None	ORR-149	0.05	392	86,700	121,200	30.0	33.8	47.7
731	8	None	ORR-149	0.05	1200	45,900	75,000	14.6	14.9	16.7
733	8	None	ORR-149	0.002	1200	48,200	64,100	6.6	6.8	11.3
732	8	None	ORR-149	0.05	1562	38,700	41,400	1.7	2.0	2.7
734	8	None	ORR-149	0.002	1562	27,500	27,500	1.0	1.3	8.9
738	8	a	ORR-149	0.05	75	108,900	134,800	27.5	30.5	11.7
739	8	a	ORR-149	0.05	392	88,400	117,900	25.4	26.4	31.9
740	8	a	ORR-149	0.05	1200	41,800	70,500	13.5	13.8	14.5
742	8	a	ORR-149	0.002	1200	42,700	61,400	9.0	9.2	9.2
741	8	a	ORR-149	0.05	1562			2.8	3.0	20.4
743	8	a	ORR-149	0.002	1562	27,900	28,000	1.1	1.6	6.0

[a]8 hr at 1600°F.

Table A-3. Creep-Rupture Properties of Welds at 1200°F in the Unirradiated Condition

Sample Number	Test Number	Weld Number	Postweld Anneal	Stress (psi)	Rupture Life (hr)	Minimum Creep Rate (%/hr)	Fracture Strain (%)	Reduction in Area (%)
35	3998	1	None	55,000	4.6	0.15	3.10	2.71
1	3780	1	None	40,000	18.7	0.013	13.1	0.77
33	5476	1	None	20,000	3864.6	0.0013	4.7	5.1
3	4080	1	None	20,000	1215.6	0.001	9.4	0.32
44	5583	1	None	20,000	3184.3	0.00080	3.6	1.9
2	3779	1	None	30,000	80.9	0.005	4.7	0.79
31	3996	1	a	40,000	308.8	0.029	19.4	13.2
30	5202	1	a	40,000	551.9	0.028	28.0	9.4
3345	5889	3	None	55,000	26.5	0.11	6.0	6.4
3345	5890	3	None	47,000	77.9	0.046	6.2	1.8
3346	5891	3	None	35,000	468.9	0.0072	7.4	1.8
	4016	3	None	40,000	273.7	0.0163	7.8	8.2
	5110	3	None	30,000	1228.0	0.0028	5.2	3.6
3395	5893	4	None	47,000	6.2	0.024	3.0	1.5
2	3997	4	None	40,000	25.2	0.0265	3.1	1.6
3	4013	4	None	30,000	80.2	0.0098	4.7	0
	5894	4	None	21,500	495.9	0.0012	1.7	0.7
7060	5140	6	None	55,000	2.8	0.584	6.3	6.4
	5881	6	None	47,000	6.3	0.0388	2.0	0.8
7061	5141	6	None	40,000	14.5	0.0326	3.2	3.2
1158	5882	6	None	30,000	158.2	0.0046	1.9	0.8
7127	5462	6	a	40,000	423.7	0.028	17.0	4.5
7071	5195	7	None	55,000	7.3	0.11	3.1	7.1
1168	5883	7	None	47,000	27.9	0.0406	3.6	1.0
7069	5192	7	None	40,000	47.8	0.016	3.1	0.8
7129	5464	7	None	32,400	179.0	0.070	2.5	0.6
7164	5312	7	None	30,000	194.8	0.0050	2.9	2.4
7135	5472	7	None	21,500	2173.2	0.0030	3.4	2.1
7126	5464	7	a	40,000	372.0	0.030	18.3	3.2
7073	5221	8	None	55,000	19.1	0.080	4.4	4.8
3340	5884	8	None	47,000	26.5	0.048	3.0	0.8
7070	5194	8	None	40,000	89.4	0.014	3.1	1.6
7125	5460	8	None	30,000	520.9	0.0035	3.5	4.0
7190	5576	8	None	21,500	3172.2	0.0012	5.2	4.0
7128	5463	8	a	40,000	692.9	0.026	29.9	23.1

[a] 8 hr at 1600°F.

Table A-4. Creep-Rupture Properties of Welds at 1200°F in the Irradiated Condition

Sample Number	Test Number	Weld Number	Postweld Anneal	Experiment Number	Stress (psi)	Rupture Life (hr)	Minimum Creep Rate (%/hr)	Fracture Strain (%)	Reduction in Area (%)
471	R-90	3	None	ETR-41-30	32,400	175.5	0.0064	1.59	
706	R-367	4	None	ORR-149	39,800	7.6	0.037	0.38	3.5
705	R-363	4	None	ORR-149	32,400	47.8	0.0037	0.20	4.3
459	R-83	4	None	ETR-41-30	32,400	111.3	0.0031	0.59	
463	R-71	4	a	ETR-41-30	32,400	585.2	0.0056	3.52	
720	R-107	6	None	ORR-149	35,000	9.6	0.020	0.37	
721	R-370	6	None	ORR-149	30,000	51.4	0.0045	0.39	12.7
535	R-377	7	None	ORR-149	39,800	17.7	0.0085	0.29	
536	R-116	7	None	ORR-149	32,400	45.1	0.0084	0.38	
735	R-108	8	None	ORR-149	35,000	3.8	0.0605	0.37	
736	R-365	8	None	ORR-149	30,000	210.8	0.0032	0.87	4.5

[a] 8 hr at 1600°F.

Table A-5. Influence of Postweld Heat Treatment on the Creep Properties of Weld 1 at 1200°F and 40,000 psi

Sample Number	Test Number	Postweld Treatment: Temperature (°F)	Postweld Treatment: Time (hr)	Postweld Treatment: Environment	Rupture Life (hr)	Minimum Creep Rate (%/hr)	Fracture Strain (%)	Reduction in Area (%)
1	3780				18.7	0.0125	3.1	0.77
22	3979	1200	100	Ar	20.3	0.0265	3.1	0.78
24	3958	1200	300	Ar	25.3	0.0557	9.4	2.4
27	5028	1200	1000	Ar	79.7	0.027	6.3	1.6
26	5112	1200	1000	Ar	39.8	0.027	9.4	1.6
42	5111	1400	2	Ar	46.6	0.014	4.7	2.4
21	3984	1400	8	Ar	97.3	0.020	4.7	3.6
19	3976	1400	20	Ar	77.6	0.032	6.3	5.1
32	3972	1400	50	Ar	320.9	0.023	12.5	7.9
34	4024	1400	200	Ar	296.6	0.030	15.0	17.2
45	5315	1400	1000	Ar	434.3	0.024	15.6	14.6
	3996	1600	0.5	Ar	54.5	0.023	3.1	4.0
	3989	1600	2	H_2	200.0	0.023	9.4	6.4
4	3793	1600	2	Ar	193.9	0.029	9.4	8.6
7	3817	1600	8	Ar	416.8	0.033	21.9	11.0
31	3966	1600	8	H_2	308.8	0.029	19.4	13.2
30	5202	1600	8	H_2	551.9	0.028	28.0	9.5
37	3995	1600	50	Ar	151.7	0.032	6.3	7.3
38	4065	1600	50	Ar	193.6	0.046	12.5	9.4
43	5249	1600	50	Ar	266.3	0.033	15.6	21.3

INTERNAL DISTRIBUTION

1–3. Central Research Library
4–5. ORNL — Y-12 Technical Library
Document Reference Section
6–15. Laboratory Records Department
16. Laboratory Records, ORNL RC
17. ORNL Patent Office
18. R. K. Adams
19. G. M. Adamson, Jr.
20. R. G. Affel
21. J. L. Anderson
22. R. F. Apple
23. C. F. Baes
24. J. M. Baker
25. S. J. Ball
26. C. E. Bamberger
27. C. J. Barton
28. H. F. Bauman
29. S. E. Beall
30. R. L. Beatty
31. M. J. Bell
32. M. Bender
33. C. E. Bettis
34. E. S. Bettis
35. D. S. Billington
36. R. E. Blanco
37. F. F. Blankenship
38. J. O. Blomeke
39. E. E. Bloom
40. R. Blumberg
41. E. G. Bohlmann
42. C. J. Borkowski
43. G. E. Boyd
44. J. Braunstein
45. M. A. Bredig
46. R. B. Briggs
47. H. R. Bronstein
48. G. D. Brunton
49–53. D. A. Canonico
54. S. Cantor
55. W. L. Carter
56. G. I. Cathers
57. O. B. Cavin
58. J. M. Chandler
59. F. H. Clark
60. W. R. Cobb
61. H. D. Cochran
62. C. W. Collins
63. E. L. Compere
64. K. V. Cook
65. W. H. Cook
66. L. T. Corbin
67. B. Cox
68. J. L. Crowley
69. F. L. Culler
70. D. R. Cuneo
71. J. E. Cunningham
72. J. M. Dale
73. D. G. Davis
74. W. W. Davis
75. R. J. DeBakker
76. J. H. DeVan
77. S. J. Ditto
78. A. S. Dworkin
79. I. T. Dudley
80. D. A. Dyslin
81. W. P. Eatherly
82. J. R. Engel
83. E. P. Epler
84. D. E. Ferguson
85. L. M. Ferris
86. A. P. Fraas
87. H. A. Friedman
88. J. H Frye, Jr.
89. W. K. Furlong
90. C. H. Gabbard
91. R. B. Gallaher
92. R. E. Gehlbach
93. J. H. Gibbons
94. L. O. Gilpatrick
95. G. M. Goodwin
96. W. R. Grimes
97. A. G. Grindell
98. R. W. Gunkel
99. R. H. Guymon
100. J. P. Hammond
101. B. A. Hannaford
102. P. H. Harley
103. D. G. Harman
104. W. O. Harms
105. C. S. Harrill
106. P. N. Haubenreich
107. R. E. Helms

108. P. G. Herndon
109. D. N. Hess
110. J. R. Hightower
111–113. M. R. Hill
114. H. W. Hoffman
115. D. K. Holmes
116. P. P. Holz
117. R. W. Horton
118. A. Houtzeel
119. T. L. Hudson
120. W. R. Huntley
121. H. Inouye
122. W. H. Jordan
123. P. R. Kasten
124. R. J. Kedl
125. M. T. Kelley
126. M. J. Kelly
127. C. R. Kennedy
128. T. W. Kerlin
129. H. T. Kerr
130. J. J. Keyes
131. D. V. Kiplinger
132. S. S. Kirslis
133. J. W. Koger
134. R. B. Korsmeyer
135. A. I. Krakoviak
136. T. S. Kress
137. J. W. Krewson
138. C. E. Lamb
139. J. A. Lane
140. C. E. Larson
141. J. J. Lawrence
142. M. S. Lin
143. R. B. Lindauer
144. A. P. Litman
145. G. H. Llewellyn
146. E. L. Long, Jr.
147. A. L. Lotts
148. M. I. Lundin
149. R. N. Lyon
150. R. L. Macklin
151. H. G. MacPherson
152. R. E. MacPherson
153. J. C. Mailen
154. D. L. Manning
155. C. D. Martin
156. W. R. Martin
157. H. V. Mateer
158. T. H. Mauney
159. H. McClain
160. R. W. McClung
161–165. H. E. McCoy
166. D. L. McElroy
167. C. K. McGlothlan
168. C. J. McHargue
169. L. E. McNeese
170. J. R. McWherter
171. H. J. Metz
172. A. S. Meyer
173. R. L. Moore
174. D. M. Moulton
175. T. W. Mueller
176. H. A. Nelms
177. H. H. Nichol
178. J. P. Nichols
179. E. L. Nicholson
180. E. D. Nogueira
181. L. C. Oakes
182. P. Patriarca
183. A. M. Perry
184. T. W. Pickel
185. H. B. Piper
186. B. E. Prince
187. G. L. Ragan
188. J. L. Redford
189. M. Richardson
190. G. D. Robbins
191. R. C. Robertson
192. W. C. Robinson
193. K. A. Romberger
194. R. G. Ross
195. H. C. Savage
196. W. F. Schaffer
197. C. E. Schilling
198. D. Scott
199. J. L. Scott
200. H. E. Seagren
201. C. E. Sessions
202. J. H. Shaffer
203. W. H. Sides
204. G. M. Slaughter
205. A. N. Smith
206. F. J. Smith
207. G. P. Smith
208. O. L. Smith
209. P. G. Smith
210. I. Spiewak
211. R. C. Steffy
212. W. C. Stoddart
213. H. H. Stone
214. R. A. Strehlow
215. D. A. Sundberg

216. J. R. Tallackson
217. E. H. Taylor
218. W. Terry
219. R. E. Thoma
220. P. F. Thomason
221. L. M. Toth
222. D. B. Trauger
223. W. E. Unger
224. R. D. Waddell
225. G. M. Watson
226. J. S. Watson
227. H. L. Watts
228. C. F. Weaver
229. B. H. Webster
230. A. M. Weinberg
231. J. R. Weir
232. W. J. Werner
233. K. M. West
234. M. E. Whatley
235. J. C. White
236. R. P. Wichner
237. F. W. Wiffen
238. L. V. Wilson
239. J. W. Woods
240. Gale Young
241. H. C. Young
242. J. P. Young
243. E. L. Youngblood
244. F. C. Zapp

EXTERNAL DISTRIBUTION

245. G. G. Allaria, Atomics International
246. J. G. Asquith, Atomics International
247. D. F. Cope, RDT, SSR, AEC, Oak Ridge National Laboratory
248. G. W. Cunningham, AEC, Washington
249. C. B. Deering, AEC, OSR, Oak Ridge National Laboratory
250. H. M. Dieckamp, Atomics International
251. A. Giambusso, AEC, Washington
252. F. D. Haines, AEC, Washington
253. C. E. Johnson, AEC, Washington
254. W. L. Kitterman, AEC, Washington
255. W. J. Larkin, AEC, Oak Ridge Operations
256. T. W. McIntosh, AEC, Washington
257. A. B. Martin, Atomics International
258. D. G. Mason, Atomics International
259. C. L. Matthews, RDT, OSR, AEC, Oak Ridge National Laboratory
260. G. W. Meyers, Atomics International
261. J. Moteff, General Electric, Cincinnati
262. D. E. Reardon, AEC, Canoga Park Area Office
263. H. M. Roth, AEC, Oak Ridge Operations
264. M. Shaw, AEC, Washington
265. J. M. Simmons, Division of Reactor Development and Technology, AEC, Washington
266. W. L. Smalley, AEC, Washington
267. S. R. Stamp, AEC, Canoga Park Area Office
268. E. E. Stansbury, the University of Tennessee
269. D. K. Stevens, AEC, Washington
270. R. F. Sweek, AEC, Washington
271. A. Taboada, AEC, Washington
272. R. F. Wilson, Atomics International
273. Laboratory and University Division, AEC, Oak Ridge Operations

274–288. Division of Technical Information Extension