CENTRAL RESEARCH LIBRARY

LOCKHEED MARTIN ENERGY RESEARCH LIBRARIES

3 4456 0515395 4

ORNL-4781

Copy 79

THE BEHAVIOR OF WELDED JOINTS IN STAINLESS AND ALLOY STEELS AT ELEVATED TEMPERATURES

R. G. Gilliland

OAK RIDGE NATIONAL LABORATORY
CENTRAL RESEARCH LIBRARY
DOCUMENT COLLECTION
LIBRARY LOAN COPY
DO NOT TRANSFER TO ANOTHER PERSON
If you wish someone else to see this document, send in name with document and the library will arrange a loan.
UCN-7969
(3 3-67)

OAK RIDGE NATIONAL LABORATORY
OPERATED BY UNION CARBIDE CORPORATION • FOR THE U.S. ATOMIC ENERGY COMMISSION

Printed in the United States of America. Available from
National Technical Information Service
U.S. Department of Commerce
5285 Port Royal Road, Springfield, Virginia 22151
Price: Printed Copy $3.00; Microfiche $0.95

This report was prepared as an account of work sponsored by the United States Government. Neither the United States nor the United States Atomic Energy Commission, nor any of their employees, nor any of their contractors, subcontractors, or their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness or usefulness of any information, apparatus, product or process disclosed, or represents that its use would not infringe privately owned rights.

ORNL-4781

UC-80 — Reactor Technology

Contract No. W-7405-eng-26

Reactor Division

THE BEHAVIOR OF WELDED JOINTS IN STAINLESS AND ALLOY STEELS AT ELEVATED TEMPERATURES

R. G. Gilliland

AUGUST 1972

OAK RIDGE NATIONAL LABORATORY
Oak Ridge, Tennessee 37830
operated by
UNION CARBIDE CORPORATION
for the
U.S. ATOMIC ENERGY COMMISSION

LOCKHEED MARTIN ENERGY RESEARCH LIBRARIES
3 4456 0515395 4

CONTENTS

	Page
PREFACE	v
ABSTRACT	1
1. INTRODUCTION	1
2. RESULTS OF SURVEY	4
3. DISCUSSION OF COLLECTED DATA	43
4. CONCLUSIONS AND RECOMMENDATIONS	45
REFERENCES	48

PREFACE

This document is one of a series of interpretive, or state-of-the-art, reports prepared as a part of the Oak Ridge National Laboratory program entitled "High-Temperature Structural Design Methods for LMFBR Components." This program has as its goal the development of a verified design technology applicable to the high-temperature, long-term operating conditions expected for LMFBR vessels, components, and core structures.

In addition to contributing to the establishment of the overall state of the art of elevated-temperature design technology, these reports are intended to assist in identifying and interpreting near-term needs in their respective areas. They will also contribute to the identification and recommendation of potential paths of approach to some of the longer-range needs associated with the generation of a verified technology.

Professor R. G. Gilliland, of the Materials Engineering Department of the University of Wisconsin at Milwaukee, was chosen for the task of preparing this report on the behavior of weldments at elevated temperatures because of his recognized expertise and experience in the area of welding engineering and because of his familiarity with welding in the nuclear field. He is currently a consultant to the U.S. Atomic Energy Commission, Division of Compliance, on metallurgical matters, primarily those involving weldments, in nuclear plants.

This report was prepared under Contract No. AT-(40-1)-4076 between the United States Atomic Energy Commission and the University of Wisconsin, Milwaukee.

J. M. Corum
Oak Ridge National Laboratory

THE BEHAVIOR OF WELDED JOINTS IN STAINLESS AND ALLOY STEELS AT ELEVATED TEMPERATURES

R. G. Gilliland

ABSTRACT

An extensive survey was conducted of the available literature concerning the mechanical properties of weldments in types 304 and 316 stainless steel and in 2 1/4 Cr—1 Mo low-alloy steel. The temperature range of interest was 900 to 1300°F. Data on weldments in these materials were found to be sparse, especially for the case of 304 stainless steel. This situation was especially true, for all materials, in the case of cyclic data (fatigue, etc.). Comparisons of weld-metal/base-metal properties and property variations between filler metals are presented. These data are evaluated with regard to service and fabrication application and experience, and the effect of service-oriented microstructural changes on component integrity is discussed. Finally, recommendations are presented as to additional data and investigations needed to more fully understand the behavior of welded joints in stainless and alloy steels at elevated temperatures.

1. INTRODUCTION

The behavior of weldments or weld regions is an important consideration and one about which relatively little is known by most designers and stress analysts. The critical nature of weld regions (particularly the fusion-line and heat-affected-zone areas) is most apparent in situations of high deformation-imposed strain concentrations. Tests have reportedly produced failures in weld regions (between nozzles and vessels, for example) at strains far less than those which would be expected, based on the properties of the unaffected base material.[1,2] Although this potential problem is recognized, design criteria codes often assume that the strength and ductility of the weld and heat-affected zone are equal to or greater than those of the base metal; if tests show reduced values, then the stress limits of weld regions must be appropriately reduced. The recently published ASME Boiler and Pressure Vessel Code High-Temperature Nuclear Vessel Code Case 1331-5 approximately accounts for reduced weldment ductility by limiting strains in weld regions to one-half the values permitted

for parent material. These approaches largely ignore the complex nature of the problem. The following survey report describes the initial effort in an attempt to remedy this situation.

This survey of the known behavior of weldments at elevated temperatures was initiated to supplement the Oak Ridge National Laboratory (ORNL) program charged with developing a high-temperature structural design technology for liquid-metal fast breeder reactor (LMFBR) vessels, components, and core structurals, including both design methods and associated design criteria. The near-term ORNL effort and emphasis are on evaluating and supporting interim high-temperature design methods and criteria that are needed for the AEC's Fast Flux Test Facility (FFTF) and LMFBR demonstration plants. Ultimately, however, criteria must be based on an accurate knowledge of time- and temperature-dependent stress and deformation behavior throughout a component. Thus it seemed quite important that a review be conducted of the available knowledge associated with weldment properties pertaining to the particular areas defined by the ORNL program.

Therefore a state-of-the-art survey of the existing literature pertaining to the behavior of weldments at elevated temperatures was conducted and is presented here. The specific materials of primary interest were types 304 and 316 stainless steels and 2 1/4 Cr—1 Mo low-alloy (Croloy) steel. Of the three, 304 stainless steel has been emphasized and is considered most important for two reasons: (1) the FFTF uses 304 primarily, and (2) the ORNL program uses 304 throughout. The other materials have been treated in a relative manner. The temperature range of interest has been that proposed for LMFBRs (900 to 1300°F), and consideration has always been given to the fact that sodium is the primary coolant.

The data found in this survey generally are associated with one of four different welding processes: (1) gas tungsten-arc, (2) gas metal-arc, (3) shielded metal-arc, and (4) submerged-arc. Each process uses consumable electrodes except the gas tungsten-arc welding method. Processes 1 and 2 utilize an inert gas (argon, helium, carbon dioxide) to sustain (ionize) the arc and protect the electrode and weld metal from atmospheric contamination during welding. These functions are performed by slags or fluxes in the shielded metal-arc and submerged-arc processes. These fluxes

are also used to clean or deoxidize the weld metal and, in some cases, add alloying elements to the weld pool.

Choice of process generally is dependent on the application (joint design, fabrication schedule, etc.); however, some general comments as to advantages and disadvantages can be stated.[3] The gas processes usually produce cleaner welds, especially with respect to inclusions, but these processes obviously require portable gas supplies for field use. Although no gas is needed in submerged-arc welding, a handling system for the granulated flux is required. The shielded metal-arc process is most applicable to field work, because its flux is provided as a coating around the consumable electrode. The flux in both the latter processes is, however, often a source of weld-metal inclusions. Generally speaking, the gas tungsten-arc and shielded metal-arc processes are low-penetration, slow-welding methods, while the gas metal-arc and submerged-arc processes are deep-penetration, fast-welding methods.

Weld filler metals are characterized by their increased alloy content with respect to the base metal on which they will be used. This higher alloy content is needed to provide a match between the as-deposited filler metal and base metal as to composition, integrity, strength properties, and metallurgical behavior. Element loss by oxidation and control of hot cracking during solidification are only a few of the reasons for this higher alloy content required in filler metals. As the reader will note, type 308 is the designation for the filler metal used on type 304 base metal. Type 316 and 2 1/4—1 Mo Croloy base metals use filler metals having designations similar to the base metal.

Included in this survey and evaluation are such weldment properties as (1) short-time strength and ductility; (2) creep and creep rupture behavior, including ductility; and (3) fatigue, with fatigue strain as a factor. The effects of welding variables and techniques on the other properties are discussed, with a related treatment of the time-dependent metallurgical effects. The effects of irradiation on high-temperature weldment properties are included, and structural implications are emphasized.

The review includes literature published from 1926 to the present in the applicable technical society journals (ASM, ASTM, ASME, AIME, AWS,

etc.) — both foreign and domestic — as well as that by government agencies. In addition, significant information was obtained by private communication with individual investigators.

In summary, this survey effort attempted to:

1. locate and summarize all available data concerned with the materials of interest and their properties within the parameters stated;
2. present summarizations in a manner such that meaningful comparisons can be made between materials included in the survey;
3. analyze and interpret the data for quality and usefulness;
4. determine the existing need resulting from deficiencies or absences of data;
5. suggest types of new data for design applications considered potentially useful for inelastic design utilization.

2. RESULTS OF SURVEY

The significant and useful data available in the open literature for types 304 and 316 stainless steels and 2 1/4 Cr—1 Mo Croloy steels are presented in Tables 1 through 4. Stress-rupture data for these materials are presented in Table 1, stress-rupture ductilities are presented in Table 2, elevated-temperature tensile and yield strengths are presented in Table 3, and elevated-temperature tensile ductilities are presented in Table 4. These weld properties represent the open-literature data considered by the author to be significant and of high quality. Data from all-weld-metal specimens as well as from transverse, or across-the-weld, specimens are included. It should be pointed out that transverse specimens always include the three weld regions (base metal, weld metal, and heat-affected zones). Occasionally, data were found from specimens containing only the heat-affected-zone material. The different types of test specimens can be located in the tables if one notes that all-weld-metal samples do not indicate failure location, whereas across-the-weld specimens do. These tables also include any postweld or preweld heat treatment given to the specimen. The data include specimens from welds produced using either submerged-arc, gas tungsten-arc, or gas metal-arc welding processes, but

Table 1. Stress-rupture data for types 304 and 316 stainless and Croloy (2 1/4 Cr—1 Mo) steel weldments

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
4		Croloy	1350°F, 1 hr, FC	1650°F, 1 hr 50°F/hr		1100	159	20.0
							257	18.0
							639	15.0
							982	13.0
	Croloy				3		217	20.0
					3		530	17.0
					3		921	13.0
			1350°F, 2 hr	1650°F, FC	2	1000	69	33.0
					2		170	30.0
					2		619	25.0
					2		52	20.0
					2		331	15.0
					2		471	13.0
					2		1222	11.0
					2		136	30.0
					2		679	25.0
					2		104	20.0
					2		941	15.0
	Croloy Casting		1350°F, 2 hr	1650°F, FC for wrought 1725°F, AC for cast	1	1000	250	25.0
					1	1000	1110	22.0
					1	1100	44	20.0
					1		389	15.0
					1		1011	13.0
			1200°F, 8 hr, FC	Normalized, drawn	4	1100	233	25.0
					2	1200	159	15.0
			Annealed 900-930°C	None		1100	220	22.4

[a]When base metal is blank, test is all-weld-metal type.

[b]FC - furnace cooled; AC - air cooled; WQ - water quenched.

[c]1 - Base metal; 2 - weld metal; 3 - fusion line; 4 - heat-affected zone.

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
4	Croloy Casting	Croloy	Annealed 900-930°C	None	2	1100	1200	17.9
							3570	13.4
							23,500	9.0
						1200	17	15.7
							158	11.1
							755	9.0
							980	7.8
							1400	6.7
							3060	5.6
							4800	4.5
	Croloy		Annealed 900-930°C	None	2	1000	1600	22.4
					2	1000	9300	17.9
					2	1100	135	17.9
					2		330	15.7
					2		770	13.4
					2		2450	11.2
					2		4900	9.0
			None	None	1	1000	497	24.0
					1		1894	21.0
					1		2206	20.0
					1	1200	796	9.0
					4		1225	7.6
					1		2114	6.0
			None	None	1	1000	423	45.0
							561	40.0
							784	35.0
							855	32.0
							1931	27.0
							3098	25.0
							316	45.0
							503	40.0
							3770	25.0
			1350°F, 1 hr, furnace cool				1118	25.0
							2134	20.0

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
4	Croloy	Croloy	1350°F, 1 hr, furnace cool	None	1	1100	61	28.0
			None				99	25.0
							218	20.0
							522	18.0
							217	15.0
							1441	10.0
							3840	7.5
							5162	7.0
							32	20.0
							313	17.5
							487	15.0
							683	13.5
							1055	12.0
							2151	10.0
							315	20.0
							946	15.0
							1398	13.0
						1000	11	50.0
							12	45.0
						1050	17	40.0
							26	35.0
							62	30.0
			1350°F, 2 hr, FC	None		1100	19	30.0
							50	25.0
							22	20.0
							89	15.0
			1350°F, 2 hr				202	12.0
						1000	110	30.0
							1080	26.0
							1583	22.0

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
4	Croloy	Croloy	1350°F, 2 hr	None	1	1000	2040	10.0
							2040	6.5
						1100	96	20.0
							134	18.0
							1493	15.0
							1200	10.0
							2088	6.0
			Stress relieved at 1350°F			1000	31.5	35.0
							104	30.0
							390	25.0
							1450	20.0
							5500	18.0
			1725°F, air cool + 1350°F				3.5	35.0
							35	30.0
							197	25.0
							660	22.0
							2800	18.0
			1575°F, furnace cool				24	35.0
							53	30.0
							278	25.0
							800	22.0
			1350°F, 1 hr, FC			1100	5	28.0
							10	25.0
							102	20.0
							512	15.0
5	Croloy	Croloy	Stress relieved	None		900	443	80.0
							349	75.0
							1107	70.0
							2970	65.0
							3534	55.0
							3921	50.0
							6557	45.0
							1078	60.0
							2460	55.0
							7150	50.0

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
5	Croloy	Croloy	Stress relieved	None	1	900	2650	65.0
							2862	60.0
							4001	55.0
							6764	50.0
							502	65.0
							1335	60.0
							3541	55.0
							7802	50.0
							19	75.0
							174	65.0
							1730	55.0
							2697	55.0
4		308	None	None		1200	76	25.0
							174	22.0
							250	20.0
							1112	18.0
							3537	15.0
			None	None			8	28.0
							1071	14.0
							2728	12.0
							6037	10.0
							11,250	8.0
						1050	41	40.0
							260	35.0
							466	30.0
							6934	22.0
							6146	22.0
			None	None			155	35.0
							699	30.0
							2289	25.0
							5336	22.0
						1200	34	25.0
							779	15.0
							3087	11.0
							5929	9.0
							11,299	7.5

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
4		308	None	None	1	1200	100	21.0
							100	24.0
							1100	19.5
							1000	14.2
							1000	18.0
							1000	14.0
							10,000	7.8
							10,000	13.0
							10,000	9.0
						1050	100	36.0
							100	35.0
		316	1200°F, 2 hr, AC	None	1	1200	67	24.0
			1550°F, 2 hr, AC	None			100	24.0
							168	24.0
			None	None		1200	9.3	35.0
							34	30.0
							83	25.0
							158	24.0
							295	20.0
			1950°F, 1/2 hr, WQ	None		1200	126	25.0
			1200°F, 2 hr, AC				87	25.0
							124	25.0
							110	25.0
			1550°F, 2 hr, AC				107	25.0
							80	25.0
			None	None			186	25.0
							706	20.0
							2828	15.0
							5708	12.5
							8941	10.5
			None	None		1200	9	30.0
							241	20.0
							2379	15.0
							2582	14.23
							3300	12.5
							5192	10.0
	316			2000°F, 30 min, AC	2	1200	3	30.0

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment	Preweld heat treatment	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
4	316	316	None	2000°F, 30 min, AC	2	1200	50	25.0
					2		280	22.5
					2		1450	21.5
			None	None			71	32.0
							1438	25.0
							3230	24.0
							2464	22.0
							1920	17.5
							6000	12.5
			None	None		1100	1344	40.0
							6024	35.0
6	316	316	Solution treatment 1350°C	None	2	1200	100	32.0
					2		1000	10.0
							1000	13.0
					2		10,000	5.0
					2		10,000	8.0
					2		10,000	7.0
					2	1300	100	10.0
					2		100	12.0
					2		1000	8.0
					2		1000	7.0
					2	1100	1000	15.0
7	316	16-8-2[d]	None	None	4	1100(N)[e]	65.3	55.0
					4		116.4	50.0
					4		183.5	45.0
					4		171.9	40.0
					4		356.5	40.0
					4		19,800 DF[f]	35.0
					4		8,376 DF	30.0
			1100°F, 2 hr, 1925°F, 2 hr, AC	None	4		33.0	55.0

[d] 16 Cr–8 Ni–2 Mo–bal Fe.

[e] Notched test.

[f] Did not fail - removed from test.

Table 1 (continued)

Reference	Base metal[a]	Filler metal	Postweld heat treatment[b]	Preweld heat treatment[b]	Failure location[c]	Test temperature (°F)	Rupture time (hr)	Rupture stress (psi)
								x 1000
7	316	16-8-2	1100°F, 2 hr, 1925°F, 2 hr, AC	None	4	1100(N)	95.7	50.0
					4		180.4	45.0
					4		629.0	40.0
					4		1187.3	40.0
					4		10,608 DF	35.0
8	316	308	None reported	None reported	4	1050	100	36.0
							1000	28.0
							10,000	20.0
						1200	100	21.0
							1000	14.2
							10,000	7.8
		308 LC[g]				1050	100	35.5
							1000	27.0
							10,000	20.5
						1200	100	19.5
							1000	14.0
							10,000	9.0
		316 LC				1200	100	22.0
							1000	16.5
							10,000	7.7
						1350	100	12.5
							1000	7.2
							10,000	3.1
4		308	None	None	1	1050	1000	28.0
							1000	27.0
							10,000	20.0
							10,000	20.5
4		316	None	None		1200	14	30.0
							52	25.0
							145	23.0
			1950°F, 1/2 hr, WQ	None			69	24.0
							58	24.0
			1200°F, 2 hr, AC	None			119	24.0

[g]LC - Low carbon (<0.03%).

Table 2. Stress-rupture ductility of weldments in types 304 and 316 stainless steel and Croloy (2 1/4 Cr–1 Mo) steel

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Rupture time (hr)	Rupture elongation (%)
5	Croloy	Croloy	Quench, temper	None	1	900	502	15.0
							1335	13.0
							3541	6.0
							7802	5.0
							19	17.0
							174	18.0
							1730	11.0
							2697	12.0
4			None	None		1000	423	60.0
							561	72.0
							784	32.0
							855	37.5
							1931	19.0
							3098	13.0
							316	13.0
							503	28.0
							3770	16.0
			1350°F, 1 hr, FC	None			1118	26.0
							2134	50.1
			None	None		1100	61	11.8
							99	19.0
							218	15.0
			1350°F, 1 hr, FC			1100	217	19.4
							1441	9.5
							3840	5.0
							5162	8.0
			None	None		1100	32	52.0
							313	14.3
							487	20.0

[a]FC - furnace cooled; AC - air cooled; WQ - water quenched.

[b]1 - All-weld-metal; 2 - across-the-weld.

Table 2 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Rupture time (hr)	Rupture elongation (%)
4	Croloy	Croloy	None	None	1	1100	683	14.0
							1055	15.0
							2151	16.0
							315	26.0
							946	33.0
							1398	36.0
5	Croloy	Croloy	Stress relieved	None	1	900	443	2.0
							349	1.0
							2970	7.0
							3534	12.0
							3921	15.0
			Quench, tempered				6557	17.0
							1078	19.0
							2460	18.0
							7150	13.0
			Stress relieved				2650	3.0
							2862	3.0
							4001	4.0
							6764	5.0
4	Croloy	Croloy	1350°F, 2 hr, FC	None	1	1000	11	12.0
							12	16.0
						1100	19	21.0
							50	22.0
						1200	22	27.0
							89	14.0
			1350°F, 2 hr				202	9.0
					1	1000	110	38.0
							1080	37.0
							1583	29.0
						1100	96	20.0
							134	21.0
							1493	6.0
			Stress relieved 1350°F	None	1	1000	31.5	29.0
							104	29.0
							390	34.0
							1450	30.0
							5500	16.5

Table 2 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[b]	Type type[b]	Test temperature (°F)	Rupture time (hr)	Rupture elongation (%)
4	Croloy	Croloy	1750°F, AC + 1350°F	None	1	1000	35	43.0
							197	43.5
							660	39.0
							2800	34.5
			1575°F, FC				24	31.0
							53	32.5
							278	44.5
							800	52.0
			1350°F, 1 hr, FC		1	1100	10	26.1
							102	38.0
							512	21.0
			1350°F, 1 hr, FC			1100	159	43.0
							257	29.0
							639	41.0
							982	33.0
			1350°F, 2 hr,	1650°F, FC	2	1000	69	19.0
							170	17.0
							619	23.0
						1100	52	29.0
							331	13.0
							471	15.0
							1222	9.0
						1000	136	18.0
							679	30.0
						1100	104	30.0
							941	10.0
			1650°F, FC	1350°F, 2 hr		1000	250	40.0
							1110	33.0
						1100	44	38.0
							389	12.0
							1011	10.0
8	304	308 LC	None	None	1	1050	6934.3	3.5
						1200	6037.2	3.0
4	304	304	None	None	1	1200	76	23.6
							174	19.0

Table 2 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Rupture time (hr)	Rupture elongation (%)
4	304	304	None	None	1	1200	250	1.7
							1112	18.1
							3537	12.0
			None	None	1	1050	41	17.0
							260	15.0
							466	5.5
							6934	9.0
							6146	3.5
						1200	1071	4.5
							2728	.5
							6037	3.0
			None	None	1	1050	155	15.0
							669	7.0
							2289	4.5
							5336	.5
			None	None	1	1200	34	17.0
							779	6.0
							3087	1.0
8	316	316	None	None	1	1200	5192.2	4.5
6	316	316	Solution treated, 2465°F	None	2	1200	100	8.0
							100	18.0
							100	27.0
							1000	3.0
							1000	5.0
							1000	12.0
							10,000	5.0
							10,000	10.0
							10,000	15.0
						1300	100	17.0
							1000	13.0
							1000	26.0
							10,000	20.0
							10,000	17.0
							10,000	15.0
						1100	1000	10.0
							10,000	7.0
4	316	316	None	None	1	1200	14	7.0

Table 2 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Rupture time (hr)	Rupture elongation (%)
4	316	316	None	None	1	1200	52	7.0
							145	7.0
			1950°F, 1/2 hr, WQ	None			69	11.5
							58	7.5
			1200°F, 2 hr, AC	None			119	8.5
							67	6.5
			1500°F, 2 hr, AC				100	12.5
							168	16.0
			None			1200	34	30.0
							83	24.0
							158	23.0
							295	20.0
			1950°F, 1/2 hr, WQ	None			126	34.5
							87	20.5
			1200°F, 2 hr, AC				124	28.0
							110	23.0
			1550°F, 2 hr, AC				107	24.5
							80	19.0
			None	None		1200	186	17.5
							706	6.0
							2828	2.0
							5708	1.0
							8941	.5
			None		1	1200	71	12.5
							3230	10.0
							2464	20.0
			None	None		1200	241	2.5
							2379	4.5
							2582	2.0
							3300	1.5
							5192	4.5

Table 3. Elevated-temperature tensile and yield strengths of weldments in types 304 and 316 stainless steel and Croloy (2 1/4 Cr–1 Mo) steel

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Yield strength (psi)	Ultimate tensile (psi)
								x 1000
9	Croloy	Croloy	1325°F stress relief, FC	1740°F normalize 1325°F temper, FC	2	1022		60.0
			1240°F normalize 1324°F temper	None		1022		48.0
			1740°F normalize 1325°F temper	None	1	1022		45.5
			1325°F stress relief, FC	None	2	1022		62.5
			1325°F stress relief, FC	None	1	1022		48.5
10	Croloy	Croloy	1 hr at 1250°F	None	1	950	56.5	78.7
						1050	52.5	65.4
						1150	33.9	48.9
11	Croloy	Croloy	1362°C, 2 hr, air cool	None	1	1022		36.4
						1022		56.0
						1112		46.0
						1112		45.5
12	Croloy	Croloy	Normalized at 1790°F	Normalized and tempered	1	1202		30.0
						1292		15.0
			Solution treated 1350°C, quenched			1202		25.0

[a] FC - furnace cooled; AC - air cooled; WQ - water quenched.

[b] 1 - All-weld-metal; 2 - across-the-weld; X - irradiated.

Table 3 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Yield strength (psi)	Ultimate tensile (psi)
								x 1000
12	Croloy	Croloy	Solution treated 1350°C, quenched	Normalized and tempered	1	1292		13.0
4	Croloy	Croloy	None	350°F preheat	1	900	73.0	107.75
						1000	71.0	96.75
						1100	69.0	74.0
			1350°F, 1 hr			900	60.5	66.25
						1000	54.5	56.75
						1100	47.0	48.0
			1350°F, 2 hr	400°F preheat		900	51.7	65.6
						1000	41.2	52.0
						1050	37.1	47.5
						1100	29.6	42.6
			1325°F	600°F preheat		1050	48.2	57.5
						1050	50.4	60.3
						1050	54.0	64.8
						1050	43.0	56.5
			1350°F, 2 hr	1650°F, furnace cooled, 400°F preheat	2	900	30.8	60.2
						1000	29.4	52.2
						1050	27.4	45.6
						1100	26.2	44.2
						900	32.9	59.7
						1000	29.9	50.2
						1050	28.5	45.1
						1100	25.2	41.2
13	304	304	None	None	1 X	932	48.2	60.6
						1112	37.2	48.6
						1202	32.6	36.3
					1	932	35.8	54.6
						1112	35.7	43.9
						1202	30.6	32.1
					2 X	932	36.8	60.4
						1112	30.5	52.5
						1202	27.8	37.0

Table 3 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Yield strength (psi)	Ultimate tensile (psi)
								x 1000
13	304	304	None	None	2	932	30.7	58.0
						1112	32.8	50.0
14	304	308	None	None	2	1000	30.5	58.8
						1200	27.3	50.5
			Stress-relieved	None	2	1000	14.2	55.6
						1200	12.6	48.0
15	304	19-9cb	None	None		1350	25.2	38.9
4	304	304	None	400°F preheat	2	1000	26.0	53.0
						1200	25.0	44.5
	304	308				1000	37.5	53.4
						1100	36.0	48.3
						1200	30.75	40.05
						1300	27.65	34.65
16	316	316	As-welded	None	2	1200	33.0	
				1900°F, 8 hr, WQ			18.0	
				2100°F, 8 hr, WQ			19.0	
				2375°F, 8 hr, WQ			15.0	
			1900°F, 2 hr, air cool	None			11.0	
				2100°F, 8 hr, WQ			14.0	
				2375°F, 8 hr, WQ			11.0	
15	316	316	None	None	2	1350	28.1	44.7
						1350	25.7	47.2
4	316	316	None	None	1	900	40.0	64.0
						1200	32.0	45.0
				400°F preheat	2	1150	28.7	45.6

Table 4. Elevated-temperature tensile ductility of weldments in types 304 and 316 stainless steel and Croloy (2 1/4 Cr-1 Mo) steel

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Elongation (%)	Reduction in area (%)
12	Croloy	Croloy	Normalized at 1790°F	Normalized and tempered	2	1202	15.0	
						1292	23.0	19.0
			Solution treated 2465°C, quench			1202	3.0	
						1292	10.0	58.0
9	Croloy	Croloy	A-1325°F stress relief, FC	1740°F normalized 1325°F temper, FC		1022	16.0	67.0
			B-1740°F normalize 1325°F temper	None		1022	22.0	69.0
			B-1740°F normalize 1325°F temper		1	1022	38.0	71.5
			C-1325°F stress relief, FC		2	1022	15.0	67.0
			C-1325°F stress relief, FC		1	1022	36.0	71.0
10	Croloy	Croloy	1 hr at 1250°F	None	1	950	24.0	68.0
						1050	25.0	73.0
						1150	29.0	72.0
17	Croloy	Croloy	1 hr at 1250°F	1925°F normalize, 1275°F temper	1	932		50.0
						1112		47.0
						1292		85.0

[a]FC - furnace cooled; AC - air cooled; WQ - water quenched.

[b]1 - All-weld-metal; 2 - across-the-weld; X - irradiated.

Table 4 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Elongation (%)	Reduction in area (%)
4	Croloy	Croloy	None	350°F preheat	1	900	21.0	57.7
						1000	18.0	54.7
						1100	18.0	59.8
			1350°F, 1 hr			900	23.0	75.3
						1000	25.0	86.0
						1100	30.5	90.0
			1350°F, 2 hr	400°F preheat		900	16.0	54.0
						1000	26.0	59.0
						1050	22.0	51.0
						1100	23.0	45.0
			1350°F	600°F preheat		1050	24.0	75.0
						1050	23.0	68.0
						1050	22.0	72.0
						1050	20.0	67.0
			1350°F, 2 hr	1650°F, furnace cooled, 400°F preheat	2	900	16.0	
						1000	10.0	
						1050	16.0	
						1100	22.0	
				Normalized, drawn		900	24.0	
						1000	26.0	
						1050	28.0	
						1100	30.0	
11	Croloy	Croloy	740°C, 2 hr, air cool	None	1	1022	22.0	75.0
						1022	23.4	81.0
						1112	27.6	87.0
						1112	24.2	84.0
13		304	None	None	1 X	932	21.3	
						1112	23.8	
						1202	13.4	
					1	932	25.0	
						1112	21.1	
						1202	27.1	
					2 X	932	19.2	
						1112	23.0	
						1202	12.1	

Table 4 (continued)

Reference	Base metal	Filler metal	Postweld heat treatment[a]	Preweld heat treatment[a]	Test type[b]	Test temperature (°F)	Elongation (%)	Reduction in area (%)
13	Croloy	304	None	None	2	932	25.5	
						1112	21.6	
						1202	16.0	
14	304	308	None	None	2	1000	33.0	54.0
						1200	33.0	54.0
			Stress-relieved	None		1000	44.0	54.0
						1200	40.0	55.0
15	304	19-9cb	None	None		1350	18.7	70.2
4		304			1	1000	39.0	66.0
						1200	35.0	56.5
		308				1000	32.3	60.2
						1100	30.9	59.9
						1200	34.4	58.8
						1300	41.5	55.1
15	316	316	None	None	2	1350	20.0	47.6
						1350	16.0	15.6
4		316	None	200°F preheat	1	900	33.0	54.0
						1200	36.0	58.0
	316	316		2050°F, 3 hr, WQ				
				400°F preheat	2	1150	22.0	51.0

do not indicate which process was used in obtaining each particular set of data. However, the author feels that this is not necessarily important in the overall usefulness of the information. Of course, if the reader desires, the references for this material are given, and the welding process can be found.

The data shown in the tables are summarized and presented in Figs. 1 through 9. Figures 1 to 3 represent the collected data for type 304 weldments. Figure 1 shows the stress-rupture and elongation vs time data for specimens tested at 900, 1050, and 1200°F. The data shown on the stress-rupture curves represent more than ten data points for each of the temperatures shown. All data are not included as points, in an effort to reduce confusion. Note that the stress-rupture—elongation weld data are for 1050 and 1200°F only. In addition, note the random scatter of the stress-rupture elongation data and the very low elongation values in the higher-temperature—longer-time region. Figure 2 shows the yield and tensile data between 900 and 1300°F for welds in 304 stainless steel. Tensile ductility is shown in Fig. 3. Note the random and scattered nature of the data.

Figures 4 to 6 represent the compilation of data for type 316 stainless steel weldments. Figure 4 is a plot similar to Fig. 1 for the stress-rupture elongation data of this material. In this case, both notched and unnotched specimens are shown in the stress rupture curves, and the stress-rupture elongation data include all of the test temperatures shown for the stress-rupture curves in this figure. Again, note the random scatter of the stress-rupture elongation data and the reduction to very low values at high test temperatures. Figures 5 and 6 represent the tensile and yield strength and elongation data at elevated temperatures available for this material.

Figures 7 to 9 are the compilations of data for 2 1/4 Cr—1 Mo Croloy steel. The stress-rupture data and stress-rupture elongation data are for test temperatures of 900, 1000, 1100, and 1200°F. The very random scatter of the data points for stress rupture and elongation is observed again, as with the other two materials. Figure 8 represents the tensile and yield strengths of this material between 900 and 1300°F, and Fig. 9 shows the elevated-temperature tensile ductility values.

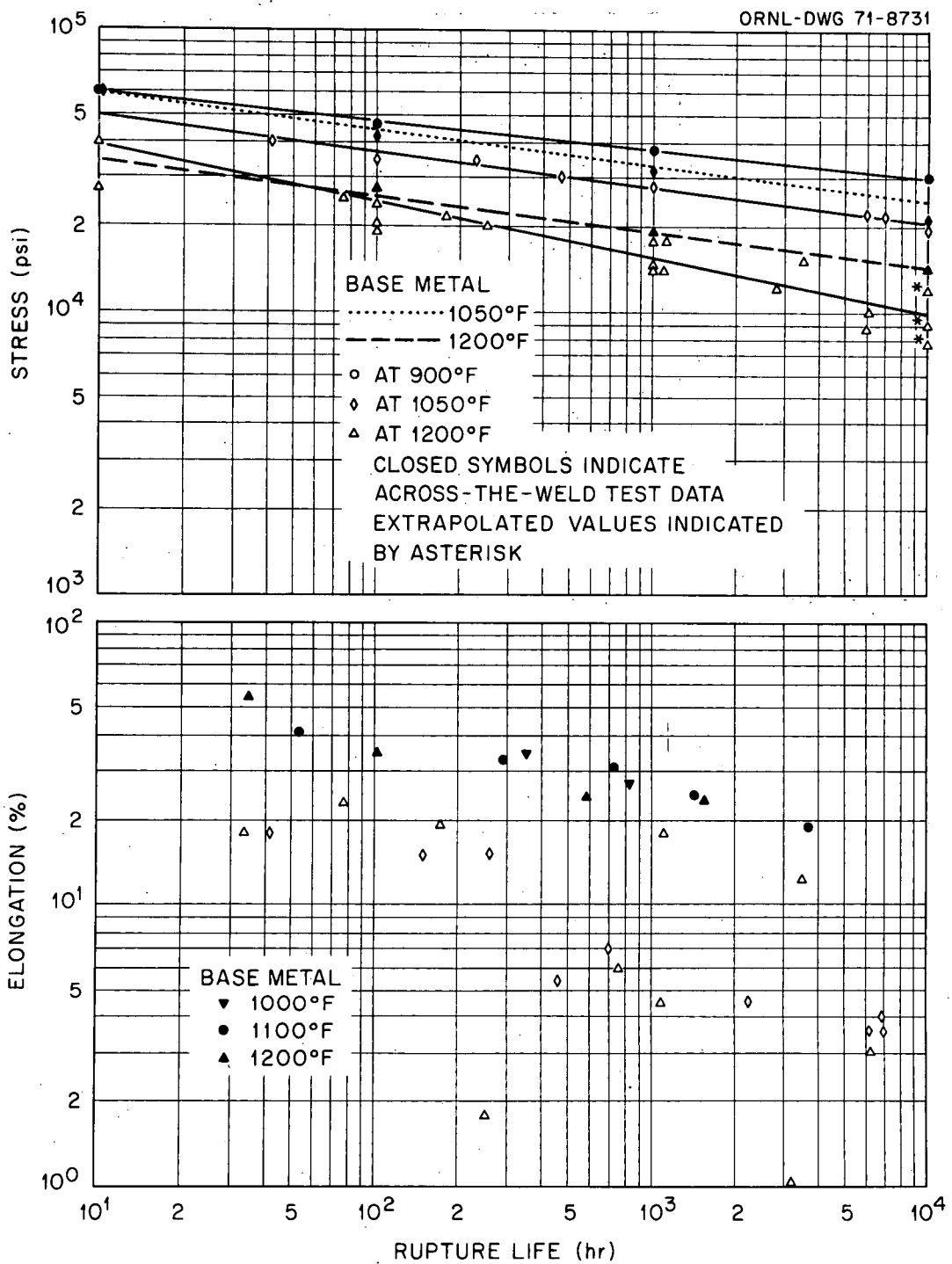

Fig. 1. Stress-rupture and elongation data for welds in type 304 stainless steel material tested at 900, 1050, and 1200°F. Creep elongation data are for tests at 1050 and 1200°F only. All-weld-metal samples were prepared using type 308 stainless steel filler metal.

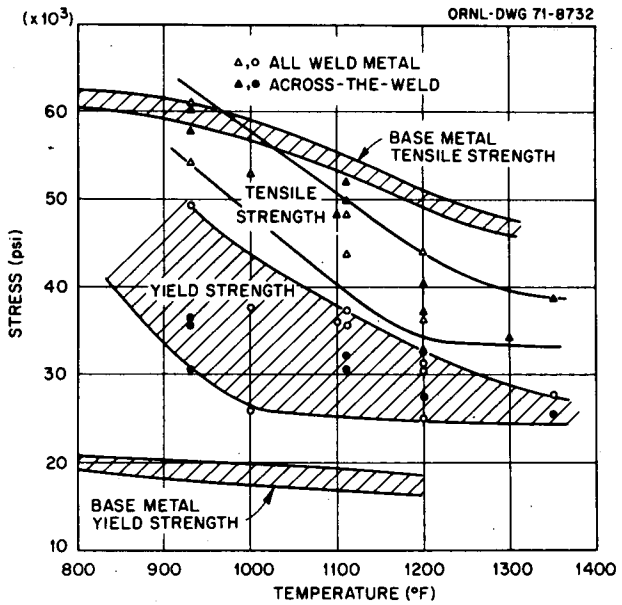

Fig. 2. Tensile and yield strength data for type 304 stainless steel welds tested between 900 and 1300°F. All-weld-metal tests were prepared using type 308 stainless filler metal.

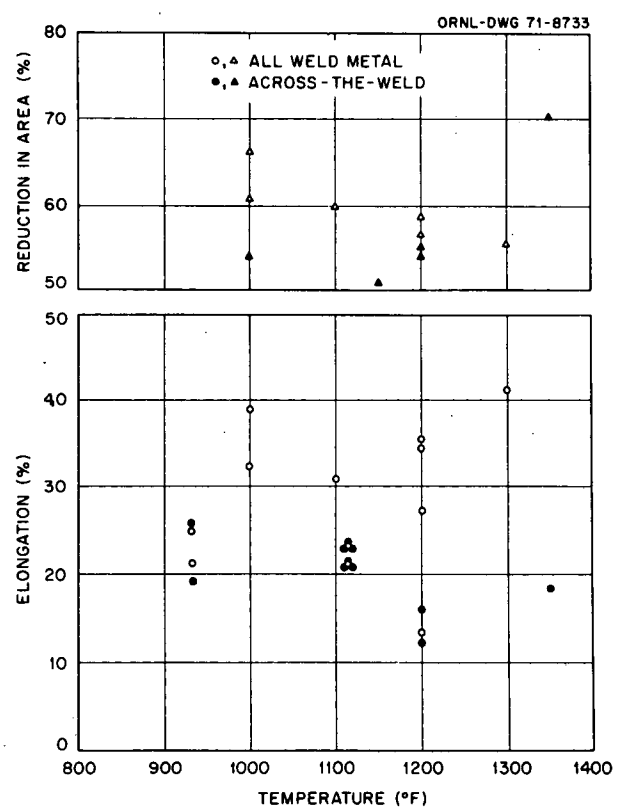

Fig. 3. Tensile elongation and reduction in area data for welds made in type 304 (308 filler) stainless steel material and tested between 900 and 1300°F.

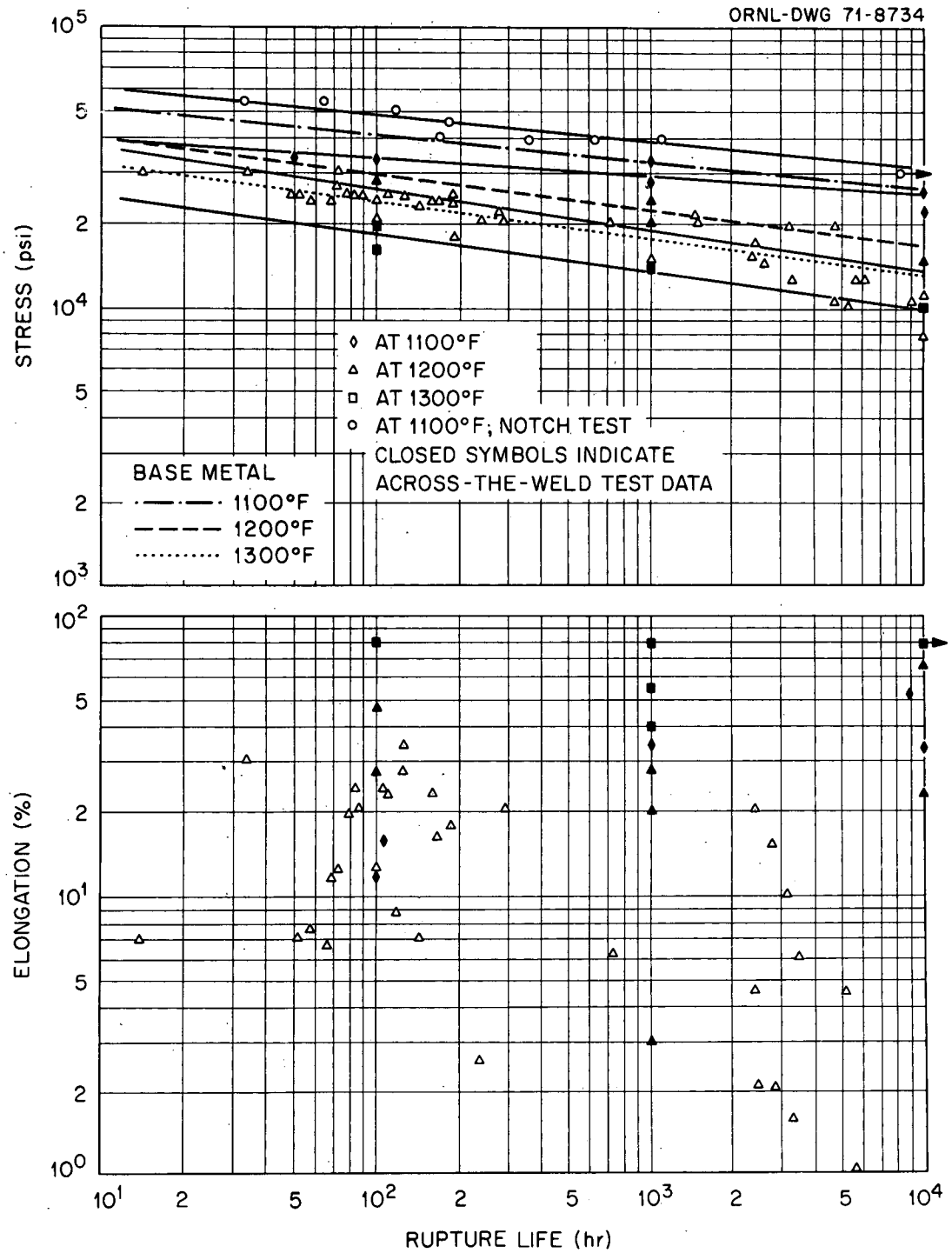

Fig. 4. Stress-rupture and elongation data for welds in type 316 stainless steel material tested at 1100, 1200, and 1300°F. Data are included from tests made using both all-weld-metal and transverse (across-the-weld) specimens (transverse data represented by closed symbols).

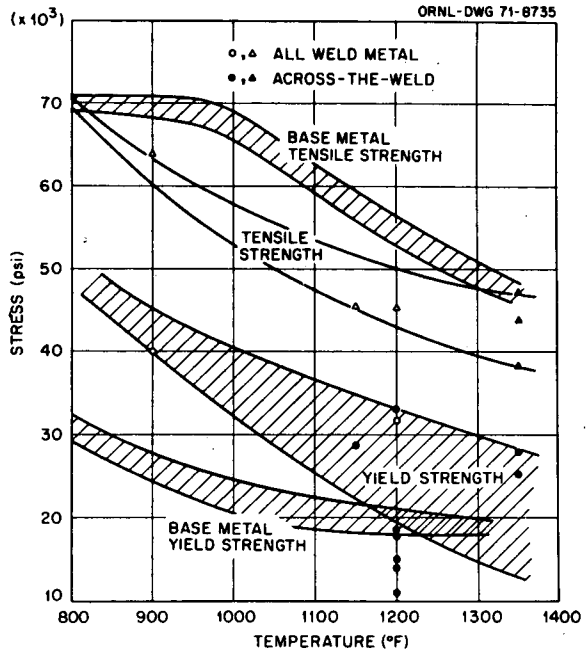

Fig. 5. Tensile and yield strength data from welds made in type 316 (316 filler) stainless steel material and tested between 900 and 1300°F.

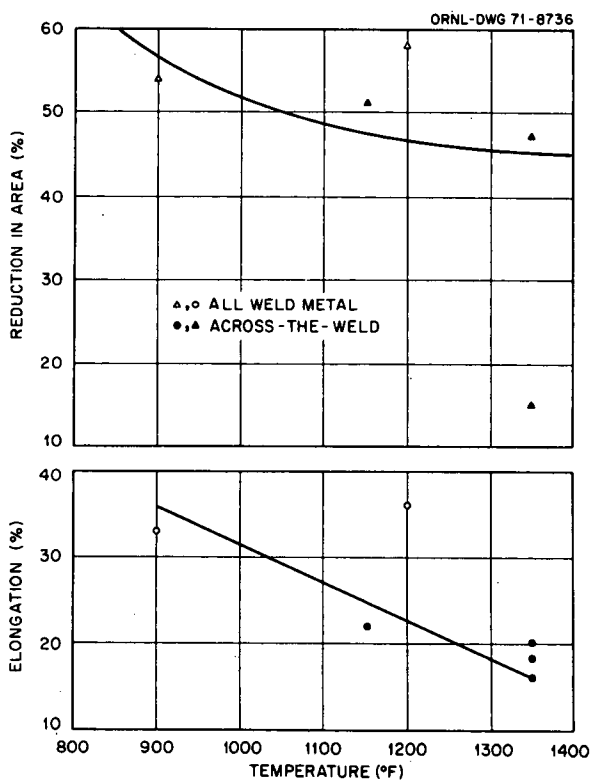

Fig. 6. Tensile elongation and reduction in area data from welds made in type 316 (316 filler) stainless steel material and tested between 900 and 1300°F.

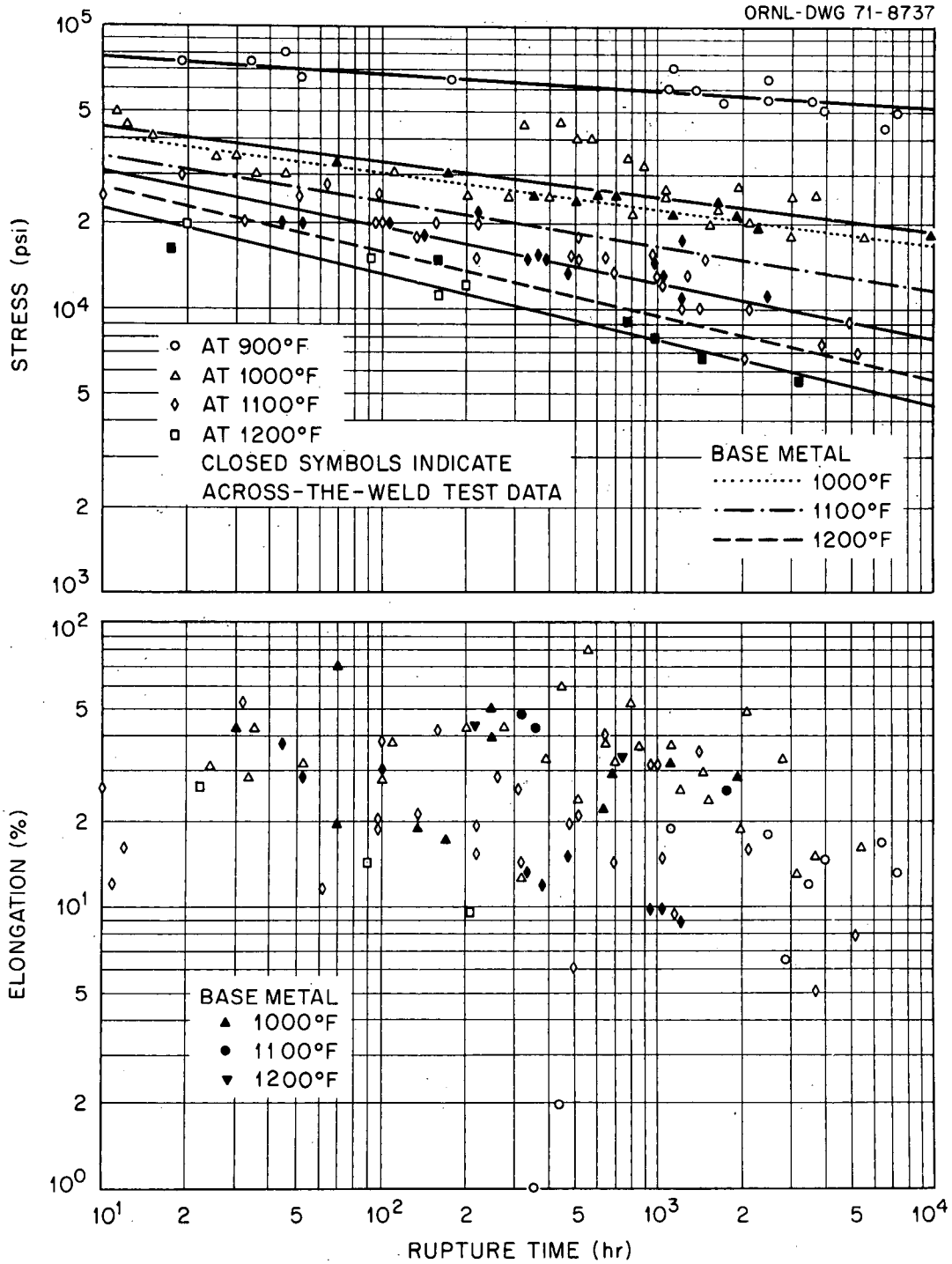

Fig. 7. Stress-rupture and elongation data for welds in 2 1/4 Cr–1 Mo steel tested at 900, 1000, 1100, and 1200°F. Data are included from tests made using both all-weld-metal and transverse specimens (transverse data designated by closed symbols).

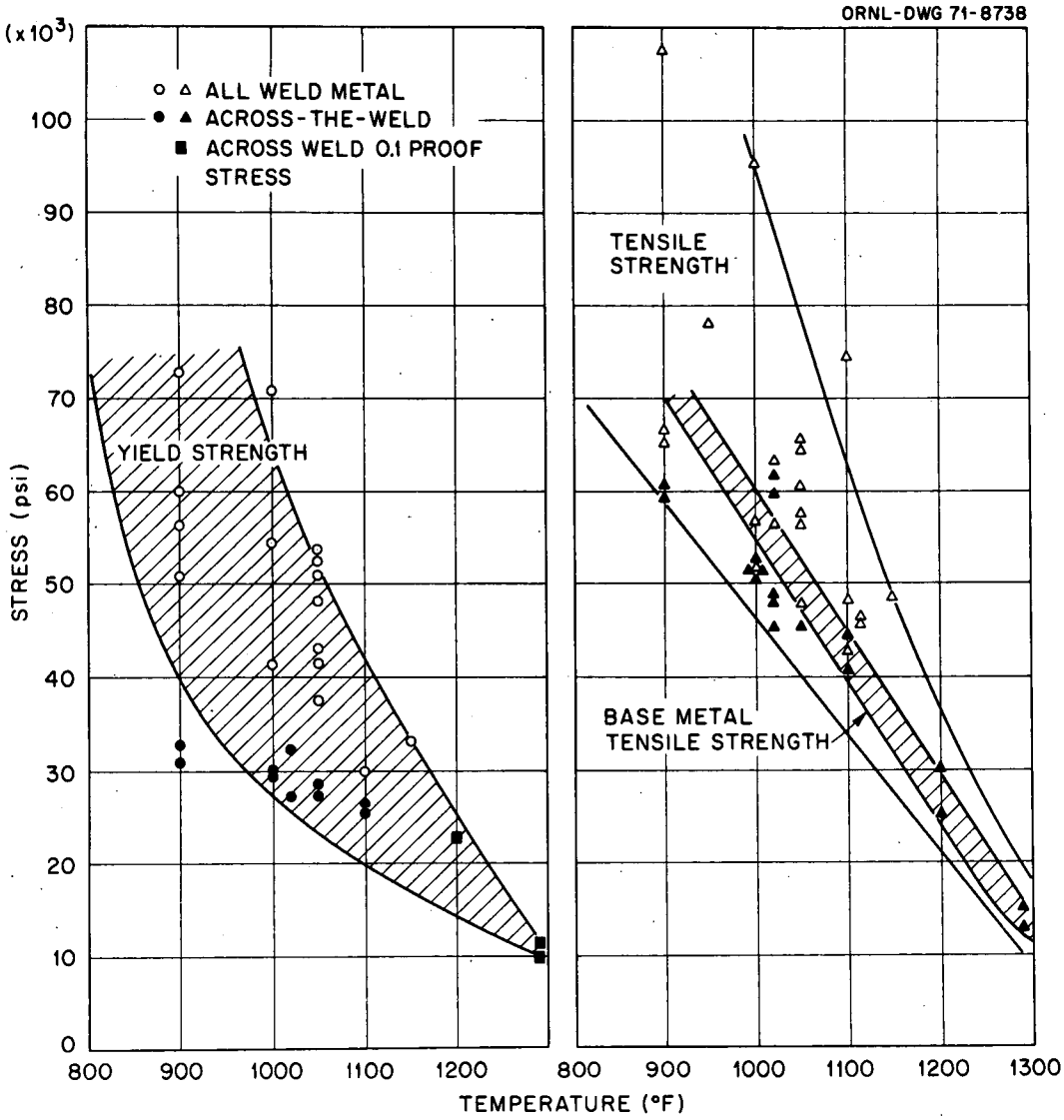

Fig. 8. Tensile and yield strength data for welds in 2 1/4 Cr—1 Mo Croloy steel tested between 900 and 1300°F. Scatter bands represent data from tests using both all-weld-metal and transverse (bar) specimens.

Note that, where possible, base-metal properties have been included in the appropriate figures discussed above (the base-metal properties were taken from Refs. 12, 18, 19, and 20). A discussion of the relative property differences in these data will be presented later.

In addition to the elevated-temperature properties reported and summarized in the tables and figures discussed above, considerable and important

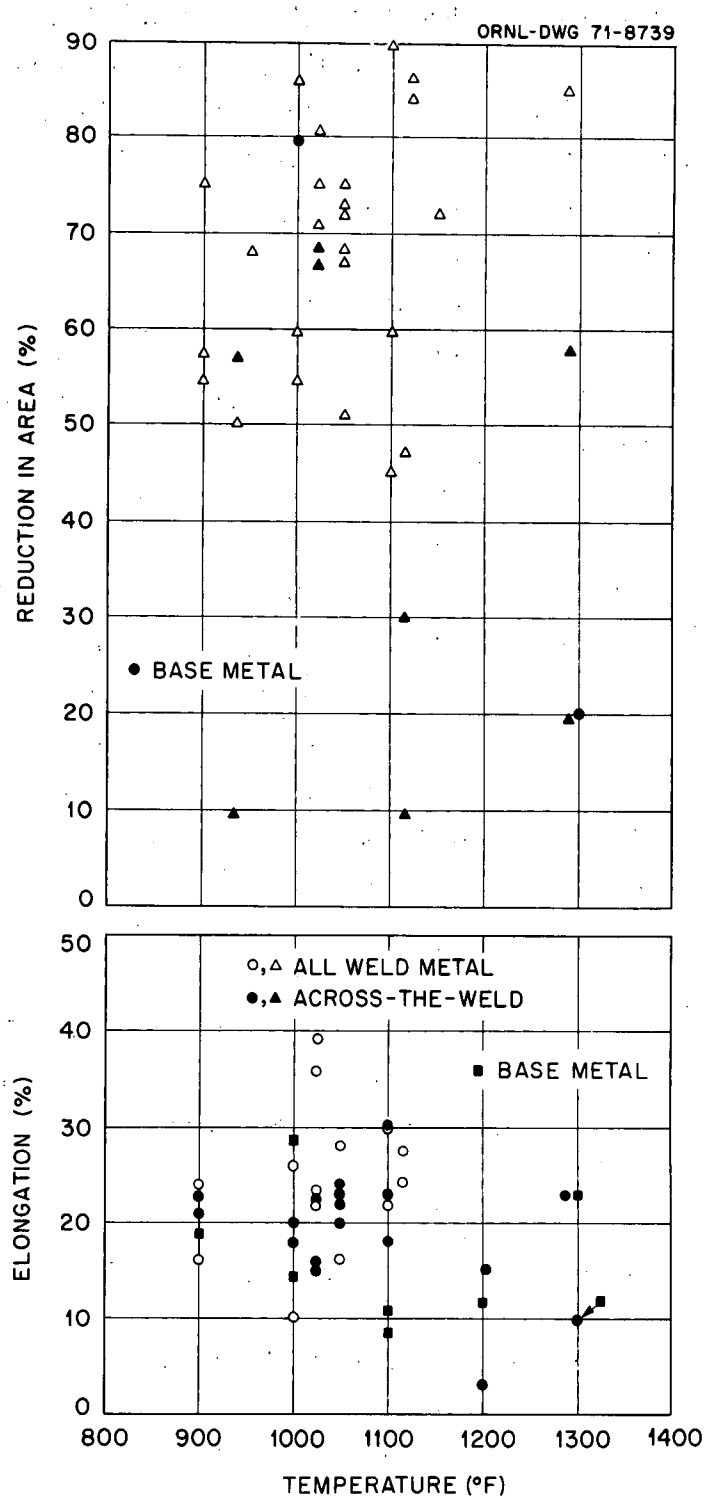

Fig. 9. Tensile elongation and reduction in area data for welds made in 2 1/4 Cr—1 Mo Croloy steel and tested between 900 and 1300°F.

information has been found in the literature which relates service experience, both in experimental and in actual applications, and metallurgical behavior of these structural steels at temperatures in the range of 900 to 1300°F. The following is a discussion of these data.

Henry et al.[21] discussed the room-temperature impact properties of weldments in austenitic stainless steels (type 304 and others) and the effect of long-time aging (15 hr minimum) at 1200, 1350, and 1500°F. The test specimens (transverse) included weld metal, the heat-affected zones, and unaffected base metal; both notched and unnotched specimens were used. The results of these tests indicated that the room-temperature properties of type 304 material are not adversely affected when subjected to these temperatures, for at least 1500 hr.

Martin and Slaughter[13] presented data describing the elevated-temperature (20 to 900°C) behavior of 304 material and Inconel 600 in the irradiated and unirradiated conditions. They found that the base metals and welds were embrittled by irradiation but to a lesser degree in stainless steel than in Inconel. In general, the welds in both materials behaved in a manner equal to or better than the corresponding base metals.

Malone[22] presented results of a study concerned with the effect of ferrite content on service at 650 to 950°F. His results indicate that austenitic-ferritic stainless weld metals of both unstabilized (316) and stabilized (347) types are subject to severe sigma embrittlement when the ferrite content exceeds 10%. Embrittlement is increased when held at 885°F for longer times. This work was addressed to overlays on 2 1/4 Cr—1 Mo alloys.*

In a paper by Kauhausen and Kautz,[23] the experiences of welding austenitic materials in Germany are reviewed. In this survey, the creep values generated in research programs were used to assess the quality of the welds. Of interest to this review are the data presented pertaining to type 316 material. The major contribution, other than the creep data, which are similar to those presented in Fig. 4, is a discussion of the effect of weld-metal ferrite content on notch impact properties. Data are

*Overlays are considered here to be a surface cladding operation similar to the present practice used for the interior surfaces of the pressure vessel in light-water reactors.

presented for 316 weld metal which shows a reduction in impact (notched) strength from 650 to about 93 ft-lb/in.^{2} after aging at 1100°F for 10,000 hr. The property was observed to be drastically reduced even after only 1000 hr aging at 1100°F, where the 140 ft-lb/in.^{2} level was reached. It is assumed, but not reported, that these are room-temperature properties obtained using the Izod test.

Papers have been reviewed relating experience and testing of the materials in question. Baker and Soldan[24] relate experiences of 15 years with heavy-wall austenitic steels for steam piping in 1050 and 1100°F central station power service. Their experience shows that 347 steel is not recommended for heavy sections; this material has been replaced by 316, which has a very good service record. Similar data are presented by Williams and Willoughby.[25] Weisberg and Soldan[26] reported on cyclic heating tests of austenitic and ferritic piping and welds representative of a main steam piping installation for a generating station. These cycles range from room temperature to 1100°F and from atmospheric pressure to 1500 psi. The austenitic materials were generally columbium-stabilized, but 2 1/4 Cr—1 Mo alloy steel was included. Their results showed that 100 cycles did not produce any cracks in these materials. The thermal cycles discussed in both of these papers are assumed to be nonuniform in nature, since the data were obtained using material taken from operating plants.

An excellent article by Isasi et al.[27] relates service experience in a reheat steam piping application using a material identified as KROMARC-58, a highly modified type 316 alloy (20 Ni—16 Cr—10 Mn—2 Mo—0.17 N—0.015 Zr—0.01 B—0.04 C). This interesting paper relates the metallurgical and microstructural changes that the material experiences after extended service periods at 1000°F. The study was conducted on failed pipe elbows that had experienced in-service cracking near weldments. The examination revealed that failure occurred intergranularly in the heat-affected zone, primarily as a result of a continuous grain boundary network of metallic carbides and nitrides. It is the author's opinion that this information is most important here to aid in illustrating the significance of carbide precipitation (which occurs most readily in the 900 to 1300°F range) in austenitic materials and the attendant property damage that can occur when this phase becomes continuous.

Data were reported by Hoke and Eberle[15] on the condition of welded structures for steam superheaters at 5000 psi and 1250°F after 14,281 hr of operation. The experimental test element included 304 and 316 stainless steel piping. This excellent paper reports room- and elevated-temperature (1350°F) yield, tensile, and ductility data for base metals and transverse weld sections before and after service. Similar data are reported for impact (Charpy V) properties and weld side-bend tests. Finally, a rather comprehensive metallographic study was conducted. The overall findings of this work indicated that no serious deterioration of properties occurred; however, weld impact properties were lowered. Changes in properties were most rapid during the first 6959 hr of service, followed by a lower rate of change thereafter. The oxidation resistance of both alloys (304 and 316) appeared to be adequate for the service of this coal-fired unit. The investigation also revealed that the nonstabilized alloys, 304 and 316, suffered no intergranular attack. Impact properties for 316 weld metal were found to be 5.5 ft-lb at room temperature and 7.3 ft-lb at 1350°F. Type 316 base-metal properties were found to be 49 ft-lb at room temperature; no 1350°F data were reported. Columbium-stabilized 19-9 filler metal was used to weld the 304 material. Impact properties of the 347 type weld metal were found to be 6 ft-lb at room temperature and about 10 ft-lb at 1350°F. The base-metal (type 304) room-temperature property was found to be 56 ft-lb. Interestingly, data on specimens that had experienced 6950 hr in the 1250°F, 5000-psi steam environment indicated that the impact properties were altered as shown in Table 5.

Table 5. Impact properties of weldments

Sample type, metal	Room-temperature tests		1350°F tests, before service
	Before service	After 6959 hr	
304 base	56	25	
304 Cb weld	6	15	10
316 base	49	18	
316 weld	5.5	2	7.3

Significant data are reported concerning the behavior of the Croloy steels at elevated temperature. Bruscato[28] reported on the shielded metal-arc weld deposits of 2 1/2 Cr—1 Mo steel tested for susceptibility to temper and creep embrittlement (a degradation in impact properties as a result of slow cooling through, or extended exposure to, the 800 to 1000°F temperature range). This study confirmed that the temper embrittlement of this material is chemistry dependent and directly related to the manganese, silicon, phosphorus, tin, antimony, and arsenic content of the weld deposit. It was shown that manganese, silicon, phosphorus, and tin were by far the largest contributors to temper embrittlement. The study included creep rupture testing at 900 and 1000°F of two materials at different impurity levels. At these temperatures, the materials containing high temper embrittlement impurities (phosphorus, tin, antimony, and arsenic) exhibit creep embrittlement, as evidenced by a deterioration in ductility properties. The material containing low impurities did not experience this deterioration. The deterioration in ductility properties of the weld deposit with a high level of temper embrittlement is particularly severe at 1050°F.

Binkley[29] has reviewed the state of the art of the weldability of stabilized 2 1/4 Cr—1 Mo steels. This review has addressed itself to the addition of titanium, niobium, vanadium, and/or tantalum to this basic steel composition in an effort to stabilize the carbon content and prevent loss of carbon to the flowing sodium in a sodium-to-water steam generator, such as in LMFBR applications. Although this report contains no numerical data as to high-temperature properties, the author feels it is worthy of being included in this review from the standpoint of its high interest to many laboratory projects currently being performed. Often, metallurgical advantages realized by base-metal carbide stabilization have resulted in poor weld-metal impact properties when using Croloy filler materials. This paper cites a commercially available alloy that includes the addition of niobium with 0.5 nickel to improve impact strength of the weld metal. Binkley discusses problems in welding that could arise primarily from weld-associated cracking due to the formation of a eutectic composition between iron and niobium. The conclusion of the paper is that since little actual experience with the alloy exists in the United States, a thorough weldability study of the niobium-stabilized Croloy is warranted.

Egnell[30] reported on the mechanical properties and weldability of a columbium-stabilized 2 1/4 Cr—1 Mo steel for LMFBR systems. Data presented in this paper indicated that solution heat treatments at higher temperatures than 1920°F strongly increased the creep rupture strength when tested at 840, 930, and 1020°F. Heat-affected zones in a weld joint have low creep ductility, as indicated by creep rupture tests on materials solution-treated at 2370°F.

A study by Swift[31] of the mechanism of stress relief cracking in quenched and tempered 2 1/4 Cr—1 Mo steel has shown the phenomenon to be closely related to precipitates that form during stress relieving at elevated temperatures. The results of the welding tests have been correlated to the Charpy V-notch toughness of the base metal. Modifications in the postweld heat-treating cycle have substantially reduced cracking. A similar study with confirming results has been reported by Mullery and Cadman.[17]

The rupture ductility of type 347 and 316 material in the temperature range of 1200 to 1400°F was investigated by Emerson and Jackson.[16] They found that the 1200°F yield strength of as-welded tubular 316 specimens was 33,000 psi (vs 73,000 psi at room temperature), but heat treatment between 1900 and 2300°F (grain size effects) reduced this differential to about 15,000 psi (by lowering the room-temperature strength). The yield strengths from the heat-treated specimens were always lower than those for as-welded specimens. The plastic ductility of welded 316 tubular specimens, given a postweld heat treatment at 1900°F, was found to be slightly greater than that of corresponding specimens in the as-welded condition. The 1200°F yield strengths of the heat-treated specimens were lower than those for the as-welded specimens.

Lupakov and Kuz'michev[32] reported on the stress rupture of 316 stainless steel at 650°C (1202°F) and 750°C (1382°F). Specimens were TIG (no filler) welded and cut from 40-mm-diam tubing (3-mm wall). Results showed that 100-hr rupture occurred at 17,000 psi at 750°C and 34,000 psi at 650°C. Elongations were of the order of 22% for 650°C and 36% for 750°C.

Lister, Mickleraith, and Higginbottom[33] and Truman and Hardwick[6] presented two excellent papers on the stress-rupture properties of 316 stainless steel. The former paper deals with steam pipe applications,

includes 2 1/4 Cr—1 Mo steels, discusses filler metals and joint designs, and reports data at 600 and 650°C. The latter paper reports only 316 data at 600 and 700°C but includes effects of heat treatment.

The creep properties of 2 1/4 Cr—1 Mo and 316 weld metal were investigated by Hopkin, Murry, and Duval.[34] Their study also included transverse specimens, with 347 being the base metal in the case of 316 weld metal. The data here agree with those previously reported. However, these experiments revealed that failure was most prevalent at the fusion line or within the heat-affected zone.

An excellent review article by Stewart and Schreitz[35] presents creep rate data for Croloy steels. Minimum creep rates of 10^{-5}, 10^{-4}, and 10^{-3}%/hr at 1000°F are reported to occur at 6300, 10,000, and 18,000 psi respectively. Similar data at 1100°F are presented, and the same minimum creep rates are found to occur at 3500, 7200, and 16,000 psi respectively. These data are presented from tests conducted on all-weld-metal samples. Fatigue data are given for columbium-stabilized stainless steel weld metal at 100 and 1100°F.

Very preliminary fatigue data for type 304 stainless steels welded with type 308 stainless steel have been obtained from ORNL.[36,37] These data are from room-temperature, 1000, and 1200°F tests using both base-metal and all-weld-metal samples, which were tested in either the as-welded or stress-relieved condition. These cursory data are presented in Fig. 10. The preliminary nature of the data is obvious from the few data points shown for the elevated-temperature curves. Although the data shown for tests at 1000 and 1200°F are insufficient for engineering application, they do allow some limited evaluation. The room-temperature tests indicated that weldments exhibit a reduction in life by a factor of 5 at high strain levels, when compared with the base metal. Even here failures occurred in the base metal. It was assumed that this behavior was the result of the higher yield strength of the weld metal. A stress relief (1850°F for 1/2 hr) seemed to improve weldment properties. Elevated-temperature vacuum tests (1000 and 1200°F) again showed that as-welded (the gas tungsten-arc process was used) weldments failed at fewer cycles than base-metal specimens; however, failure now occurs, in general, in the weld region of the specimen. It was reported that failure cracks appeared to nucleate at weld-metal

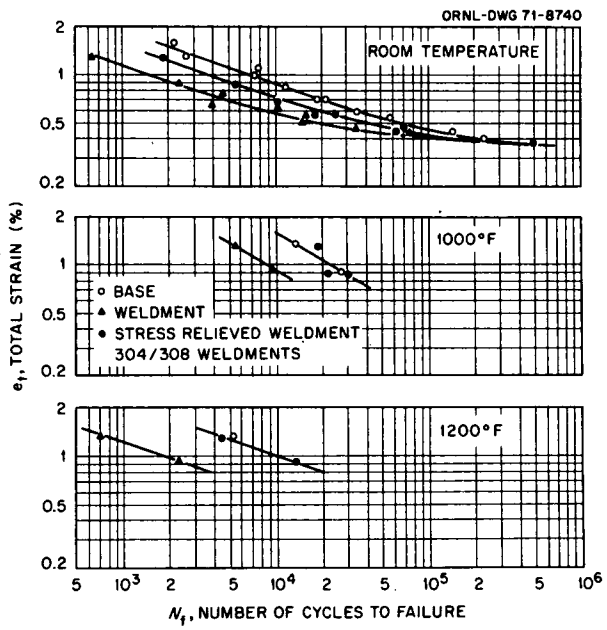

Fig. 10. Total strain-fatigue data for type 304/308 weldments.

defects, such as pores and inclusions, ultimately resulting in failure. These data are sufficient evidence that considerable effort should be expended to improve on the situation in this area.

Recently, low-cycle fatigue data have been published by workers at Combustion Engineering on type 308 weld metal.[38] These experiments were performed to determine if premature failure occurred in type 308 weld specimens compared with unwelded specimens of type 304 base material when subjected to a combined low-cycle fatigue (±1% strain) and creep test at 1050°F. The specimens included transverse weld, longitudinal weld, and base-metal types. Shielded metal-arc (with controlled residual elements) and submerged-arc type 304 welds were included in the study. The data revealed that failure always occurred at or below 200 cycles for these 1050°F strain-controlled experiments. In addition, it was concluded that submerged-arc welds appeared to be more susceptible to creep and low-cycle fatigue damage than both shielded metal-arc weld (with controlled residual elements) and the base metal. Finally, the shielded arc weld was seen to be superior to the base metal with regard to resistance to low-cycle fatigue and creep damage.

Significant information is currently being generated in programs sponsored by the AEC concerned with the elevated-temperature properties of weldments in stainless steels, particularly 304 and 316. A cooperative program among the Oak Ridge National Laboratory, Argonne National Laboratory (ANL), Hanford Engineering Development Laboratory (HEDL), and Aerojet Nuclear Company (ANC) is under way, which involves a design test matrix on one batch of FFTF welds and an evaluation of four others.[39] The following is a brief summary of the status of these programs.

The ORNL program is concerned with filler-metal composition and solidification structure and their relationship to the method of welding as applied to 304 stainless steel weldments. This program has direct application to LMFBR systems. The studies have primarily been concentrated in the area of welding using the classical stick electrode (shielded metal-arc welding) and submerged-arc techniques.[40] In all cases, filler-metal wires that conform to the standard specification for type 304 electrodes are being used; but, in most cases, compositional modifications have been made to the flux coating on either the rod or shield flux material. The data generated thus far in this program are essentially summarized in Figs. 11 to 13. Figure 11 presents the stress rupture elongation properties of shielded metal-arc welds at 1200°F. The stress vs rupture time for these welds is shown in Fig. 12. Figure 13 shows the elongation vs rupture time of welds made using the submerged-arc process, and Fig. 14 presents the stress vs rupture time data for the same specimens. Note that in all cases the tests were conducted at 1200°F. In those cases where it is appropriate, base-metal values are shown. It is obvious that the trend of these data is toward lower ductility with increasing rupture time (decreasing stress) and that at 1200°F, total elongation values of less than 1% occur frequently. Currently, data are being generated at 1100°F to confirm that similar ductility losses occur at this lower temperature. Preliminary results of tests performed using submerged-arc techniques have recently been published by Weir and Slaughter.[14] This report states that the elevated-temperature mechanical properties of the weldment depend on the solidification substructure. Since this substructure can be greatly influenced by variables within a process, the objective of the program is to determine the relationship between welding

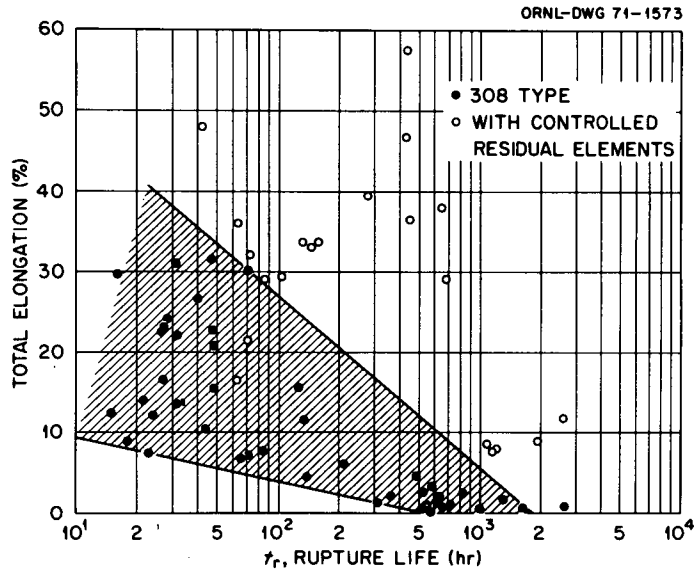

Fig. 11. Stress-rupture elongation properties of type 308 stainless steel shielded metal-arc weld deposits at 1200°F. The shaded band is for standard commercial electrode weld metal (see Fig. 1). The open points represent ORNL data from welds made using electrodes with experimental coatings that employed controlled residual elements.

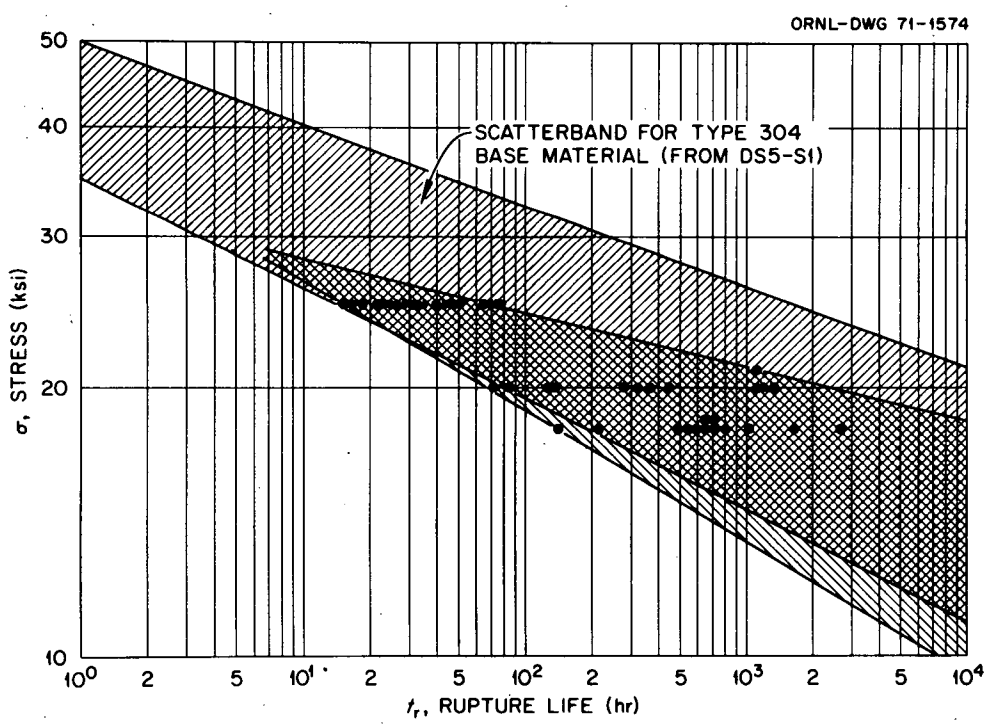

Fig. 12. Stress-rupture properties for type 308 stainless steel shielded metal-arc welds at 1200°F. The points represent data from the ORNL experimental study (see data in Fig. 1).

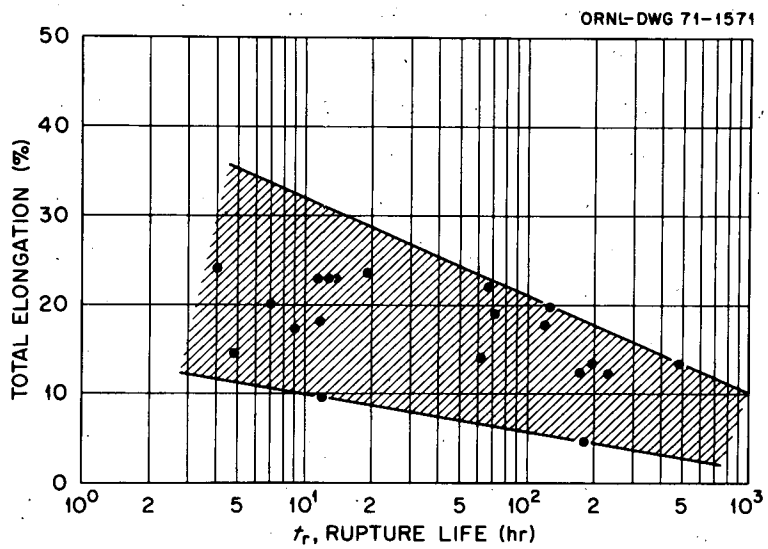

Fig. 13. Stress-rupture elongation properties of type 308 stainless steel submerged-arc weld metal at 1200°F. The data shown are from the ORNL experimental program (no commercial data are shown for comparison).

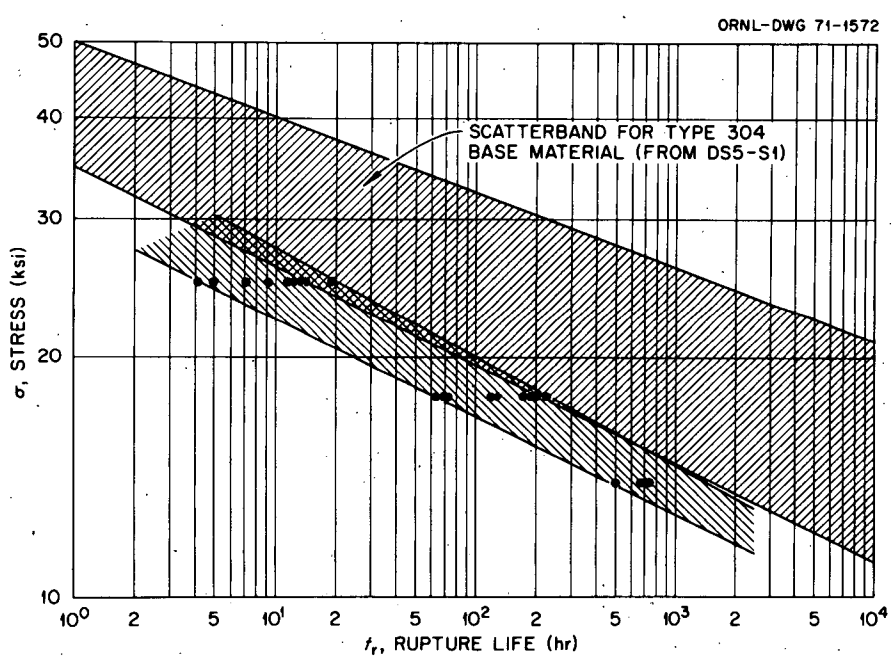

Fig. 14. Stress-rupture properties of type 308 stainless steel submerged-arc welds at 1200°F. Data points are from the ORNL program.

variables and substructure and, subsequently, the mechanical properties. Much of the program on submerged-arc weldments has progressed only to the point where unexplained variances deter the dissemination of specific data. The creep rupture properties obtained at 1100°F as a function of welding heat input and stress are not consistent. The results from shielded metal-arc weldments showed marked ductility loss with increasing time to rupture at 1200°F which was caused by metallurgical instability attributable to small composition differences. This information is an extension of that received earlier in the private communication.[40] The results of a study to examine the effect of compositional changes on creep rupture properties at 1200°F and 25 ksi indicated that boron additions to the flux material of the order of 0.004 to 0.006% showed the best combination of strength and ductility. In general, the ORNL program will include tensile tests at low strain rates, high-accuracy creep rupture, stress relaxation, and an evaluation of fatigue properties.

Extensive details of the programs at HEDL, ANL, and ANC were not available at the time of this writing. However, the following describes the general program functions to be conducted at these facilities. The Westinghouse-Hanford Laboratory will conduct high-strain-rate tensile tests on as-annealed, aged, and irradiated samples. They will perform creep rupture testing at 1000, 1100, and 1200°F, with additional concern for creep-damage evaluation. The Argonne effort will be centered on the establishment of a fatigue design test matrix, which will include the effect of hold time on these properties. The Aerojet program will concentrate on a postirradiation fatigue design test matrix.

Communication with personnel at Westinghouse Advanced Reactor Division related the essential objectives of a proposed program there that would be concerned with the high-temperature property evaluation of welded stainless steel structures.[41] It was learned that their program would primarily be concerned with: (1) fracture strain data related to the vessel life, or expected vessel life, and (2) the potential effect on these properties of metallurgical notches as manifested primarily by welds. The latter concern is a result of the work being done by Manjoine et al.[42,43] It was related that the proposed Westinghouse program would include the

use of full-size specimens, measuring up to 2 3/8 in. in thickness. The program would concern itself with tensile and stress-rupture properties of both small- and full-scale samples, with testing essentially confined to elevated temperatures (1000 to 1100°F). These tests would be performed on each region of the weld (base metal, heat-affected zone, and weld metal) at 1050 and 1100°F. The program would also include a series of stress-rupture specimens with projected failures from 1000 to 10,000 hr at 1050 and 1100°F, in both the as-fabricated and heat-treated condition. Finally, a limited number of large specimens (on the order of 2 3/8 x 2 3/8 in. cross section) having all three zones would be tested at both of the temperatures mentioned and in both the tensile and stress-rupture testing modes. This proposed program would constitute a major effort in the evaluation of stainless steel in heavier sections and the material's ultimate application to reactor vessels and would be very worth while.

3. DISCUSSION OF COLLECTED DATA

The following is a brief discussion of the data presented above and an attempt to compare the data of the three materials under consideration with themselves and with their respective base metals.

A comparison of Figs. 2, 5, and 8 shows that much lower values of tensile and yield strength are exhibited by the Croloy material than by the two stainless steel materials. The comparison of weld-metal and base-metal properties of the materials reveals, generally, only subtle changes. Type 304 base metal is superior, from a tensile strength standpoint, only at the higher test temperatures (1100 to 1300°F), while 316 base-metal tensile properties are slightly higher than the weld metal. There seems to be no differences in the Croloy base-metal and weld-metal tensile properties between 900 and 1300°F. Yield strengths of the weld metals in all alloys are superior to those of similar base metals. This is probably a result of the finer microstructure inherent in the weld metal.

Based on stress-rupture data, Croloy is superior to both 304 and 316 at 900°F. As the temperature is increased, Croloy loses ground rapidly.

These are not unexpected observations, in the author's opinion. In general, the stress-rupture strengths of the base metals are superior to those of the weld metals in all three materials. However, in no case is this degradation in properties considered extremely significant. This observation is probably best explained by recognizing that larger grain structures (lower grain boundary surface areas) exhibit higher creep properties than those of fine grain.

The stress-rupture—elongation data are too random to even draw lines or scatter bands; however, Croloy data seem to be less dispersed than those for the other two materials. The small number of rupture elongation data points for 304 is alarming, and work is obviously necessary here. Note also that the 304 data presented here are entirely from all-weld-metal samples, while Croloy and 316 data include both all-weld-metal and transverse samples. Here again it should be stated that the stress-rupture elongation properties of the base metal of each of these alloys are superior to those of the weld metal in similar materials. As mentioned above, this effect is probably due to microstructural differences.

The stress-rupture values for 304 and 316, for both base metal and weld metal, seem to depend to a certain extent on the carbon level of the material. This observation is made from an analysis of the reported literature but is not indicated or shown in the figures. This difference was more prevalent at 1200°F than at 1050°F. At 1200°F, for low carbon (0.03), these values are 1000 to 5000 psi lower than for carbon levels of 0.08. At 1050°F, no noticeable difference was observed.

While transverse specimens are included, they do not seem to behave to any great extent differently than the all-weld-metal samples. This is true even though the fracture location varies from the heat-affected zone to the fusion line to the weld metal.

Probably one of the most alarming findings of this effort was the inability to find sufficient fatigue data at temperatures of 900 and 1300°F for weld specimens in the materials of interest. The data which were found[27,36] were exceedingly preliminary in nature and only consisted of work on 304 type material and weldments. Obviously, work not only on 304 but on 316 stainless steel and the Croloys is warranted. In any case, the

meager data available indicate that the weld is inferior to the base metal at elevated temperatures (1000 and 1200°F).

Only in the low-cycle fatigue data[27] reported by Combustion Engineering was a real difference in properties observed that related to welding processes. Although there may be significant data variations from process to process, it seems risky, if not both impracticable and impossible, for this author to make such conclusions unless the reference makes such observations. However, the one reference in which such findings were stated may be very significant and further supports the need for more work.

4. CONCLUSIONS AND RECOMMENDATIONS

The discussion of data in the above section essentially points out the usefulness and quality of the data found during the survey. It is the author's opinion that most of the data presented here are reliable and represent work of high quality. What is not found is much more alarming than what is found; this is the area to which this section will be addressed.

The use of high-temperature data for design applications primarily centers on a discussion of whether or not design criteria are sought for either a yielding or nonyielding situation. It has been assumed throughout this effort that design data are sought for determining the deformation behavior of materials at elevated temperatures in reactor systems. Therefore, it appears to the author that the data most useful to a design engineer would be those that would be found in basic creep data, that is, inelastic deformation vs time at constant load (or stress) and temperature. Also, cyclic load and temperature data would, obviously, be just as important and useful. This kind of information would be as valuable to the design engineer as, say, stress-rupture data. The designer would then be provided with the more meaningful data of first- and second-stage creep plus an indication of the minimum creep rate of the material at the specific conditions of load and temperature. The only information found in this survey similar to this was that which was reported for Croloy at 1000 and 1100°F, and this paper only reported the minimum creep rate vs load.[35]

Thus, it is the author's recommendation that serious consideration be given to the generation of basic creep data such that information might be available for use related to first- and second-stage creep deformation at the subject temperatures and stress levels. In addition, the absence of fatigue and other cyclic data in welded sections of these materials should be alleviated, especially the absence of low-cycle data. The suggestion here is that more information like that from the Combustion Engineering work[27] should be generated. Comparisons could then be made with existing data for base metals, and any problems that might arise because of the differences in these data could be assessed. Thus, it is the author's recommendation that fatigue testing be included with the generation of basic creep data. This program could prove to be most interesting, since the two material properties (creep and fatigue) at the design temperatures (900 to 1300°F) could affect and interact with the metallurgical reactions known to exist in this temperature range (carbide precipitation, etc.).

As implied above, the testing of materials in fatigue and creep at temperatures between 900 and 1300°F causes stainless steels that are not stabilized with titanium or columbium to undergo metallurgical reactions known as sensitization. This phenomenon is the precipitation of a complex chromium/iron carbide such that the near volumes of the grain boundaries (material within a few microns of the grain boundary) in the material are rendered corrosion-susceptible. Data are available for environments that are proposed in the LMFBR system, and these data show that corrosion does not seem to be a problem. However, the author is most concerned about the load-carrying abilities, in fatigue and creep situations, of materials that exist for long times in these temperature ranges (900 to 1300°F), resulting in a grain boundary which contains a carbide phase that can be continuous.[27] Thus, one could speculate that this grain boundary is now brittle and would easily propagate cracking, causing possible catastrophic failure. This is another metallurgical reason for recommending the generation of creep and fatigue data in these materials in these temperature ranges at the design stress levels, particularly after extended time periods at LMFBR operating conditions. In addition, it may allow a more complete explanation of the

reasons why the 1200°F tests conducted by the ORNL group show rapid loss in ductility as 1000-hr test times are approached.

Interest in fatigue (low cycle, thermal, etc.) and creep of welded stainless steel structures also occurs in other areas of nuclear power generation, outside the LMFBR program. Recent experiences with operating reactors possessing sensitized nozzle components have indicated the low level of integrity exhibited by the stainless steel microstructure. These stainless material components have been furnace sensitized during the code-required stress relief necessary during reactor construction and vessel component fabrication. ASME reports[1,2] relate the findings and method of repair of one such failure. The cause of failure has usually been considered stress corrosion, caused by high residual stresses and environmental impurities. However, recent incidents have revealed that fatigue may be a factor, since no environmental impurities and/or excessive residual stress prevailed. Thus, there is an existing need for the proposed generation of long-time high-temperature properties in reactor component materials.

In summary, based on this review, certain major areas of information are found lacking in the current available knowledge with respect to high-temperature properties of weldments in the materials of interest. Those general areas considered by the author to constitute the most important data requirements are listed below.

1. the generation of basic creep data (deformation vs time) for weldments in the subject materials, providing first- and second-stage creep properties for the proposed temperature range and stress levels;
2. the generation of cyclic load data for weldments in the proposed materials and in the temperature range of interest, with particular emphasis on the thermally and mechanically induced low-cycle fatigue properties;
3. an evaluation of the overall effect, and interactions, occurring between time- and temperature-dependent metallurgical reactions and long-time mechanical properties (both cyclic and static) for the temperature range and stress levels of interest.

These three areas have been found to be the major categories where data are missing and which are most important for the establishment of proper and useful design methods and criteria.

REFERENCES

1. H. C. Burghard, Jr., and F. F. Lyle, Jr., Investigation of Cracking in Nuclear Reactor Primary Piping System, ASME Publication No. 71-PVP-33.
2. E. B. Norris, P. D. Watson, and R. D. Wylie, Repair of Primary Pressure System Piping in a Nuclear Power Plant, ASME Publication No. 71-PVP-50.
3. Welding Handbook, American Welding Society (sect. 2, 6th ed.), 1969.
4. H. R. Voorhees and J. W. Freeman, "Elevated Temperature Properties of Weld-Deposited Metal and Weldments," ASTM STP 226, 1958.
5. G. N. Emmanuel and W. E. Leyda, "Long-Time High Temperature Properties of Cr-Mo Weld Metal," Properties of Weldments at Elevated Temperatures, American Society for Metals, pp. 1–4 (1968).
6. R. J. Truman and D. Hardwick, "Some Effects of Heat Treatment and Welding on the Rupture Properties of Three Austenitic Steels," ibid.
7. R. J. Christoffel, "Notch Sensitivity of the Heat-Affected Zone in Type 316 Material," Welding J., pp. 25-s–28-s (January 1963).
8. R. D. Wylie, C. L. Corey, and W. E. Leyda, "Stress Rupture Properties of Some Chromium-Nickel Stainless Steel Weld Deposits," Trans. ASME 76, 1093–1106 (October 1954).
9. M. Inagaki, I. Okane, and M. Nakajima, "Influence of Heat Treatment on Creep Rupture Strength of Welded Joints of 2 1/2 Cr–1 Mo Steel," Trans. Natl. Res. Inst. Metals 8(4), 34-47 (1966).
10. J. Bland, "The Arc Welding of 2.25% Cr–1.0% Mo Alloy Steel Pipe," Welding J., 181-s–94-s (April 1956).
11. K. Kremer, "Results of Joint Long-Time Creep Rupture Tests on Weld Metal Deposits of Creep-Resistant Steels," IIW Commission II (Arc Welding), Doc. II - 455-68.
12. W. Barr and M. J. Siverns, "Hot Tensile Properties of 1/2 Cr–1/2 Mo–1/4 V and 2 1/4 Cr–1 Mo Steam Pipe Material," Brit. Welding J., pp. 238–43 (May 1967).
13. W. R. Martin and G. M. Slaughter, "Irradiation Embrittlement of Welds and Brazes at Elevated Temperatures," Welding J., pp. 385-s–91-s (September 1966).

14. J. R. Weir, Jr., and G. M. Slaughter, Welding Development for LMFBR Stainless Steel Components, ORNL-4630 (March 1971).

15. J. H. Hoke and F. Eberle, "Experimental Superheater for Steam at 2000 psi and 1250°F — Report after 14,281 Hours of Operation," Trans. ASME 79, 307–17 (1957).

16. R. W. Emerson and R. W. Jackson, "The Plastic Ductility of Austenitic Piping Containing Welded Joints at 1200°F," Welding J., pp. 89-s–104-s (February 1957).

17. F. Mullery and R. O. L. Cadman, "Cracking of Welded Joints in Ferritic Heat-Resisting Steels," Brit. Welding J., pp. 212–20 (April 1962).

18. G. V. Smith, An Evaluation of the Yield, Tensile, Creep, and Rupture Strengths of Wrought 304, 316, 321, and 347 Stainless Steels at Elevated Temperatures, ASTM Data Series DS 5S2 (Supplement to Publication DS 5, formerly STP 124), 1969.

19. G. V. Smith, An Evaluation of the Elevated Temperature Tensile and Creep-Rupture Properties of Wrought Carbon Steel, ASTM Data Series DS 11S1 (Supplement to Publication DS 11, formerly STP 180), 1970.

20. W. L. Greenstreet, J. M. Corum, and C. E. Pugh, Informal Progress Reports on Structural Design Methods for LMFBR Components for Period Ending September 30, 1970, ORNL-CF-70-10-56 (Oct. 1, 1970).

21. O. H. Henry, M. A. Cordovi, and G. D. Fischer, "Sigma Phase in Austenitic Stainless Steel Weldments," Welding J., pp. 75-s–81-s (February 1955).

22. M. O. Malone, "Sigma and 885°F Embrittlement of Chromium-Nickel Stainless Steel Weld Metals," Welding J., pp. 241-s–53-s (June 1967).

23. E. Kauhausen and H. R. Kautz, "Welding of Creep Resistant Austenitic Materials," Brit. Welding J., pp. 266–79 (May 1962).

24. R. A. Baker and H. M. Soldan, "Service Experience at 1050 and 1100°F of Piping of Austenitic Steels," Proceedings of Joint International Conference on Creep — ASTM/ASME, pp. 4-85–4-94, 1963.

25. N. T. Williams and G. Willoughby, "Laboratory Assessment of Austenitic Steels for Welded Steam Pipe Joints," Brit. Welding J., pp. 115-23 (March 1962).

26. H. Weisberg and H. M. Soldan, "Cyclic Heating Test of Main Steam Piping Materials and Welds at the Sewarem Generating Station," Trans. ASME 76, 1085–91 (October 1954).

27. J. A. Isasi, R. C. Bates, and J. Heuschkel, Analysis of Cracks in Welded Elbows, ASME Publication 71-PVP-32, 1971.

28. R. Bruscato, "Temper Embrittlement and Creep Embrittlement of 2 1/4 Cr–1 Mo Shielded Metal-Arc Weld Deposits," Welding J., 44(4), 148-s–56-s.

29. N. C. Binkley, State-of-the-Art Review – Weldability of Stabilized 2 1/4 Cr–1 Mo Steels, draft of report from Metals and Ceramics Division, Oak Ridge National Laboratory, Oak Ridge, Tenn.

30. L. Egnell, "Mechanical Properties and Weldability of a Columbium Stabilized Steel with 2 1/4% Chrome and 1% Molybdenum for Sodium Cooled Fast Reactor Systems," paper presented at ASM Materials Engineering Congress, Oct. 19–22, 1970, Cleveland, Ohio.

31. R. A. Swift, "The Mechanism of Stress Relief Cracking in 2 1/4 Cr–1 Mo Steel," Welding J., 50(5), 195-s–200-s (May 1970).

32. I. S. Lupakov and Y. S. Kuz'michev, "Strength and Intercrystalline Corrosion Resistance of Kh18N12M2T Steel Weld Joints," Metalloved. i Term. Obrabotka Metal. 10, 60–63 (1962).

33. E. Lister, J. Mickleraith, and A. Higginbottom, "High Temperature Properties of Steam Pipe Welds," Iron and Steel Institute, Publication 97, 1967, pp. 473–81.

34. L. M. T. Hopkin, D. Murray, and D. Duval, "Creep Properties of Some Butt Welds in Steampipes," J. Iron Steel Inst., pp. 819–25 (August 1965).

35. W. C. Stewart and W. G. Schreitz, "Thermal Shock and Other Comparison Tests of Ferritic Steels for Main Steam Piping – A Summary Report," Trans. ASME, pp. 1056–67 (August 1953).

36. Private communication with R. W. Swindeman, Oak Ridge National Laboratory, Metals and Ceramics Division, March 1971.

37. W. L. Greenstreet, J. M. Corum, and C. E. Pugh, Informal Progress Reports on Structural Design Methods for LMFBR Components for December 1970 and January 1971, ORNL-CF-71-2-48 (Feb. 1, 1971).

38. J. K. Hayes and D. M. Vandergriff, Test Report on Low Cycle Fatigue and Combined Creep Test of Type 308 Weld Metal, CENC-1157 (June 1971).

39. A. L. Ward et al. (WADCO Corporation) and G. M. Slaughter et al. (ORNL), Preparation, Characterization, and Testing of Austenitic Stainless Steel Weldments, HEDL-TME 71-118: UC-25 (August 1971).

40. Private communication with G. M. Goodwin, Oak Ridge National Laboratory, Metals and Ceramics Division, March 1971.

41. Private communication with F. Snow, Westinghouse Advanced Reactor Division, Waltz Mill, Pa., March 1971.

42. S. K. Chan, M. J. Manjoine, and C. Visser, Analysis of Stresses in Pressurized Welded Pipe in the Creep Range, ASME Publication No. 71-PVP-66.

43. M. J. Manjoine, "High Temperature Stress-Rupture Analysis," Scientific Paper 71-1E7-Mechs-P1, Westinghouse Research Laboratories, Pittsburgh, Pa., Mar. 9, 1971.

ORNL-4781

UC-80 — Reactor Technology

Internal Distribution

1. S. E. Beall
2. H. W. Behrman
3. E. E. Bloom
4. S. E. Bolt
5. J. W. Bryson
6. J. P. Callahan
7. H. P. Carter
8. S. J. Chang
9. R. D. Cheverton
10. C. J. Claffey
11. C. W. Collins
12-31. J. M. Corum
32. J. S. Crowell
33. F. L. Culler
34. W. G. Dodge
35. W. R. Gall
36. G. M. Goodwin
37-39. W. L. Greenstreet
40. A. G. Grindell
41. R. C. Gwaltney
42. W. O. Harms
43. R. F. Hibbs (Y-12)
44. P. R. Kasten
45. R. T. King
46. K. C. Liu
47. M. I. Lundin
48. T. S. Lundy
49. R. N. Lyon
50. R. E. MacPherson
51. W. R. Martin
52. W. J. McAfee
53. H. E. McCoy
54-56. H. C. McCurdy
57. J. G. Merkle
58. S. E. Moore
59. H. A. Nelms
60. P. Patriarca
61. T. W. Pickel
62. C. E. Pugh
63. M. Richardson
64. W. K. Sartory
65. M. R. Sheldon
66. G. M. Slaughter
67. J. E. Smith
68. D. A. Sundberg
69. R. W. Swindeman
70. D. B. Trauger
71. R. S. Valachovic
72. R. D. Waddell, Jr.
73. A. M. Weinberg
74. J. R. Weir
75. G. D. Whitman
76. G. T. Yahr
77. H. C. Young
78. ORNL Patent Office
79-81. Central Research Library
82. Document Reference Section
83-103. Laboratory Records
104. Laboratory Records — RC

External Distribution

105. L. E. Alsager, U.S. Atomic Energy Commission, Washington, D.C. 20545
106. W. F. Anderson, Liquid Metal Engineering Center, Canoga Park, Calif. 91304
107. D. L. Becker, Westinghouse Hanford Co., Richland, Wash. 99352
108. I. Berman, Foster Wheeler Corp., Livingston, N.J. 07039
109. C. C. Bigelow, U.S. Atomic Energy Commission, Washington, D.C. 20545
110. E. C. Bishop, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
111. L. D. Blackburn, Westinghouse Hanford Co., Richland, Wash. 99352
112. M. Booth, U.S. Atomic Energy Commission, Washington, D.C. 20545

113. T. Branca, Teledyne Materials Research Co., Waltham, Mass. 02154
114. C. R. Brinkman, Aerojet Nuclear Co., Idaho Falls, Idaho 83401
115. J. P. Burn, Westinghouse Hanford Co., Richland, Wash. 99352
116. D. R. Campbell, Liquid Metal Engineering Center, Canoga Park, Calif. 91304
117. A. E. Carden, University of Connecticut, Storrs, Conn. 06268
118. A. K. Chakraborty, General Electric Co., San Jose, Calif. 95125
119. S. K. Chan, Westinghouse Research Laboratory, Pittsburgh, Pa. 15235
120. T. Y. Chang, The University of Akron, Akron, Ohio 44304
121. F. L. Cho, Sargent and Lundy, Chicago, Ill. 60603
122. R. L. Citerly, ANAMET Laboratories, San Carlos, Calif. 94070
123. T. T. Claudson, Westinghouse Hanford Co., Richland, Wash. 99352
124. J. C. Cochran, Liquid Metal Engineering Center, Canoga Park, Calif. 91304
125. W. J. Collins, U.S. Atomic Energy Commission, Division of Compliance, Washington, D.C. 20545
126-127. Combustion Engineering, Inc., Chattanooga, Tenn. 37401, Attn: Librarian MR and D
128. J. B. Conway, Mar-Test, Inc., Cincinnati, Ohio 45215
129. W. E. Cooper, Teledyne Materials Research Co., Waltham, Mass. 02154
130. R. F. Detman, C. F. Braun and Co., Alhambra, Calif. 91802
131. G. S. Dhaliwal, Valcor Engineering Corp., Kenilworth, N.J. 07033
132. R. J. Dohrmann, Babcock and Wilcox Co., Alliance, Ohio 44601
133. W. D. Doty, U.S. Steel Corp., Monroeville, Pa. 15146
134. J. J. Droher, Liquid Metal Engineering Center, Canoga Park, Calif. 91304
135. J. M. Duke, Westinghouse Electric Corp., Tampa, Fla. 33616
136. A. J. Durelli, The Catholic University of America, Washington, D.C. 20017
137. T. J. Dwyer, Control Data Crop., Palo Alto, Calif. 94304
138. A. F. Eckert, Babcock and Wilcox Co., Lynchburg, Va. 24501
139. R. H. Engelken, U.S. Atomic Energy Commission, Division of Compliance, Washington, D.C. 20545
140. E. P. Esztergar, 252 Palomar St., LaJolla, Calif. 92037
141. N. J. Ettenson, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
142. L. F. Fidrych, General Electric Co., Sunnyvale, Calif. 94086
143. W. N. Findley, Brown University, Providence, R.I. 02912
144. R. H. Gallagher, Cornell University, Ithaca, N.Y. 14850
145. J. C. Gerdeen, Michigan Technology University, Houghton, Mich. 49931
146-155. R. G. Gilliland, University of Wisconsin, Milwaukee, Wis. 53201
156. G. A. Greenbaum, Mechanics Research, Inc., Los Angeles, Calif. 90045
157. Nicholas Grossman, U.S. Atomic Energy Commission, Washington, D.C. 20545
158. O. Hagen, Westinghouse Electric Corp., EMD, Cheswick, Pa. 15024
159. J. Hagstrom, Chicago Bridge and Iron Co., Oak Brook, Ill. 60523
160. G. R. Halford, NASA Lewis Research Center, Cleveland, Ohio 44135
161. E. E. Hamer, Argonne National Laboratory, Argonne, Ill. 60439

162. D. G. Harman, Westinghouse Electric Corp., Tampa, Fla. 33616
163. D. R. Harting, Boeing Co., Aerospace Group, Seattle, Wash. 98124
164. K. Hayashi, Esso Research and Engineering Co., Florham Park, N.J. 07932
165. D. B. Heilig, Bechtel Corp., San Francisco, Calif. 94119
166. K. Hikido, General Electric Co., Sunnyvale, Calif. 94086
167. F. P. Hill, Combustion Engineering, Inc., Chattanooga, Tenn. 37402
168. J. P. Houstrup, Combustion Engineering, Inc., Chattanooga, Tenn. 37401
169. J. R. Hunter, U.S. Atomic Energy Commission, Washington, D.C. 20545
170. M. T. Jakub, Westinghouse Hanford Co., Richland, Wash. 99352
171. C. E. Jaske, Battelle Memorial Institute, Columbus, Ohio 43201
172. R. I. Jetter, Atomics International, Canoga Park, Calif. 91304
173. H. B. Ketchum, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
174. E. E. Kintner, U.S. Atomic Energy Commission, Washington, D.C. 20545
175. W. C. Kinsel, Westinghouse Hanford Co., Richland, Wash. 99352
176. L. Kornblith, U.S. Atomic Energy Commission, Division of Compliance, Washington, D.C. 20545
177. R. E. Kosiba, U.S. Atomic Energy Commission, Washington, D.C. 20545
178. H. Kraus, Rensselaer Polytechnic Institute of Connecticut, Hartford, Conn. 06120
179. E. Krempl, Rensselaer Polytechnic Institute, Troy, N.Y. 12181
180. Y. S. Lai, Chicago Bridge and Iron Co., Oak Brook, Ill. 60521
181. C. F. Larson, Welding Research Council, New York, N.Y. 10017
182. C. W. Lawton, Combustion Engineering, Inc., Windsor, Conn. 06095
183. M. M. Lemcoe, Battelle Memorial Institute, Columbus, Ohio 43201
184. A. Lohmeier, Westinghouse Electric Corp., Tampa, Fla. 33616
185. Josef Maatuk, Crane Co., Chicago, Ill. 60632
186. R. H. Mallett, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
187. J. R. Maloney, Westinghouse Electric Corp., EMD, Cheswick, Pa. 15024
188. M. J. Manjoine, Westinghouse Electric Corp., Pittsburgh, Pa. 15235
189. P. V. Marcal, Brown University, Providence, R.I. 02912
190. C. B. McGough, Westinghouse Hanford Co., Richland, Wash. 99352
191. J. L. McLean, Teledyne Materials Research Co., Waltham, Mass. 02154
192. R. Mello, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
193. J. L. Mershon, U.S. Atomic Energy Commission, Washington, D.C. 20545
194. W. R. Mikesell, Chicago Bridge and Iron Co., Oak Brook, Ill. 60523
195. R. A. Moen, Westinghouse Hanford Co., Richland, Wash. 99352
196. W. I. Morita, Bechtel Corp., San Francisco, Calif. 94119
197. E. C. Norman, U.S. Atomic Energy Commission, Washington, D.C. 20545

198. E. T. Onat, Yale University, New Haven, Conn. 06520
199. W. J. O'Donnell, O'Donnell and Associates, Pittsburgh, Pa. 15232
200. D. H. Pai, Foster Wheeler Corp., Livingston, N.J. 07039
201. C. S. Parker, General Electric Co., San Jose, Calif. 95125
202. W. Pennell, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
203. A. G. Pickett, Southwest Research Institute, San Antonio, Texas 78228
204. F. C. Rally, General Electric Co., Sunnyvale, Calif. 94086
205. Y. R. Rashid, General Electric Co., San Jose, Calif. 95125
206. G. W. Reinmuth, U.S. Atomic Energy Commission, Division of Compliance, Washington, D.C. 20545
207. B. T. Resnick, U.S. Atomic Energy Commission, Washington, D.C. 20545
208. D. R. Riley, U.S. Atomic Energy Commission, Washington, D.C. 20545
209. E. C. Rodabaugh, Battelle Memorial Institute, Columbus, Ohio 43201
210. M. A. Rosen, U.S. Atomic Energy Commission, Washington, D.C. 20545
211. F. A. Ross, U.S. Atomic Energy Commission, Washington, D.C. 20545
212. Robert Rung, Applied Technology Associates, Inc., Ramsey, N.J. 07446
213. P. H. Sager, Gulf General Atomic, San Diego, Calif. 92112
214. R. I. Sann, Gulf United Nuclear Fuel Corp., Elmsford, N.Y. 10523
215. A. O. Schaefer, Metals Properties Council, New York, N.Y. 10017
216. R. A. Schoup, U.S. Atomic Energy Commission, ORO, Oak Ridge, Tenn. 37830
217. J. J. Schreiber, U.S. Atomic Energy Commission, Washington, D.C. 20545
218. C. C. Schultz, Babcock and Wilcox Co., Alliance, Ohio 44601
219. F. Sebring, C. F. Braun and Co., Alhambra, Calif. 95125
220. R. A. Selbey, Westinghouse Hanford Co., Richland, Wash. 99352
221. L. K. Severud, Westinghouse Hanford Co., Richland, Wash. 99352
222. Paul Shahinian, Code 6396, Naval Research Laboratory, Washington, D.C. 20390
223. M. Shaw, U.S. Atomic Energy Commission, Washington, D.C. 20545
224. P. G. Shewman, Argonne National Laboratories, Argonne, Ill. 60439
225. Sidney Siegel, Atomics International, Canoga Park, Calif. 91304
226. J. M. Simmons, U.S. Atomic Energy Commission, Washington, D.C. 20545
227. E. E. Sinclair, U.S. Atomic Energy Commission, Washington, D.C. 20545
228. R. J. Slember, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
229. G. V. Smith, 104 Berkshire Road, Ithaca, N.Y. 14850
230. A. L. Snow, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
231. K. O. Stein, Babcock and Wilcox Co., Alliance, Ohio 44601
232. W. H. Sutherland, Westinghouse Hanford Co., Richland, Wash. 99352
233. J. A. Swanson, 870 Pine View Drive, Elizabeth, Pa. 15037
234. A. V. A. Swaroop, Metallurgy Department U-136, University of Connecticut, Storrs, Conn. 06268
235. A. N. Tardiff, U.S. Atomic Energy Commission, Washington, D.C. 20545

236. Kevin Thomas, Westinghouse Electric Corp., ARD, Madison, Pa. 15663
237. J. C. Tobin, Westinghouse Hanford Co., Richland, Wash. 99352
238. I. S. Tuba, Basic Technology, Inc., Pittsburgh, Pa. 15235
239. W. H. Tuppeny, Combusion Engineering, Inc., Windsor, Conn. 06095
240. R. A. Valentin, Argonne National Laboratory, Argonne, Ill 60439
241. W. A. Van der Sluys, Babcock and Wilcox Co., Alliance, Ohio 44601
242. S. A. Weber, Westinghouse Hanford Co., Richland, Wash. 99352
243. R. W. Weeks, Argonne National Laboratory, Argonne, Ill. 60439
244. B. C. Wei, U.S. Atomic Energy Commission, Washington, D.C. 20545
245. D. H. White, Westinghouse Electric Corp., Tampa, Fla. 33616
246. B. V. Winkel, Idaho Nuclear Corp., Idaho Falls, Idaho 83401
247. B. M. Wundt, 2346 Shirlane Road, Schenectady, N.Y. 12309
248. S. Y. Zamrik, Pennsylvania State University, University Park, Pa. 16802
249. Zenons Zudans, Franklin Institute Research Lab., Philadelphia, Pa. 19103
250. U.S. Atomic Energy Commission Office of Regulation, Library, Washington, D.C. 20545
251-253. Director, Division of Reactor Licensing, Washington, D.C. 20545
254-255. Director, Division of Reactor Standards, Washington, D.C. 20545
256. Patent Office, AEC, ORO
257-273. Manager, Technical Information Center, Oak Ridge
274. Research and Technical Support Division, AEC, ORO
275-489. Given distribution as shown in TID-4500 under Reactor Technology category (including 25 copies to NTIS)