2018

A STUDY OF
PIPING MATERIALS
FOR SODIUM PIPING
IN LMFBR PLANTS

TECHNICAL REPORT 228

PREPARED FOR

MASTER

U S ATOMIC ENERGY COMMISSION

CONTRACT NO AT(04-3)-781

BRAUN PROJECT 4122-W

UNITED NUCLEAR PROJECT 2351

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May 22, 1970

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BAL-128

Dear Mr Fry

PIPING MATERIALS
TECHNICAL REPORT 228
LMFBR PIPING DESIGN GUIDE
PROJECT 4122-W

Transmitted herewith are two copies of the final report covering piping materials for sodium piping in LMFBR piping systems.

The purpose of the study was to conduct a state-of-the-art survey on the selected materials for sodium piping. Basic properties and data on the effects of environmental conditions were compiled for use in preparation of the piping design guide. Some additional literature data has been added since the preliminary draft report was issued for review.

Comments on the report from reviewers will be welcome but further revision is not contemplated.

Sincerely yours

Roger Detman
Project Manager

RFD JH

Project 4122-W

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PIPING MATERIALS FOR LMFBR PLANTS

Technical Report 228

(Revision 1)

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Prepared for U. S. Atomic Energy Commission

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Braun Project 4122-W

United Nuclear Project 2351

UNITED NUCLEAR CORPORATION

Research and Engineering Center

Elmsford, New York

May 22, 1970

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1. INTRODUCTION

1.1 GENERAL

C F Braun & Co, with United Nuclear Corporation as a major sub-contractor, has been awarded a contract* by the U. S. Atomic Energy Commission for the development and verification of a Piping Design Guide for sodium-cooled fast breeder reactor power plants (LMFBR). The work is a Priority 1 Task under the LMFBR Program Plan† prepared for the USAEC Division of Reactor Development and Technology (RDT) by the LMFBR Program Office, Argonne National Laboratory. It is identified in the Program Plan as Task 3-8.2, Development of Design Technology for Piping.

This Technical Report presents a state-of-the-art survey on the selected materials for sodium piping in LMFBR plants. It covers materials properties, with special emphasis on the effects of environmental conditions to which the pipe materials will be exposed. The output of this study will provide the required background for subsequent Tasks 314, 315, and 316 which are to define Materials R&D Requirements, Recommended Materials Properties, and Erosion and Corrosion Considerations, respectively. The information developed will be factored into appropriate chapters of the Piping Design Guide.

It is anticipated that the criteria developed in this Technical Report will be subject to revision and/or expansion as the work progresses. A review procedure has been established to ensure consistency between interdependent studies currently being performed under this and other task areas of the LMFBR Program Plan.

Under the review procedure, the Liquid Metal Engineering Center (LMEC) is the coordinating agency. Technical Reports prepared by Braun under this contract will be distributed by LMEC to appropriate Review Agencies designated by USAEC. Comments from Review Agencies will be evaluated by Braun to determine their validity. Valid comments then will be incorporated into subsequent revisions to the Technical Reports.

*AT(04-3)-781, AEC San Francisco Operations Office (SAN).

†Liquid Metal Fast Breeder Reactor Program Plan, LMFBR Program Office, Argonne National Laboratory, AEC R&D Report, Reactor Technology, Wash-1101 (Aug. 1968).

Fig. 1.1 shows the interrelationship of tasks under Task Area 3-8, Coolant Transport and Storage Systems.*

1.2 PURPOSE

The purpose of this study is to provide information on basic properties and data on the effects of environmental conditions on selected materials for sodium piping. Emphasis is given to those effects that are of direct concern to the piping engineer: corrosion and erosion, and changes in mechanical properties. Besides the pipe and fitting materials, weld filler materials are also considered in this study.

1.3 SELECTED MATERIALS FOR STUDY

Materials have been selected for this study under Task 100 (reported in Technical Report 110). The primary selections are types 304 and 316 stainless steels, hereinafter called 304 SS and 316 SS, and $2\frac{1}{4}\%$ Cr-1% Mo steel, hereinafter called $2\frac{1}{4}$ Cr-1 Mo. Specific compositions and grades are covered in Section 2.2.

The selected materials are widely used for high-temperature piping in the utility and process industries. An International Nickel Company stainless steel market study (Reference 1 and Private Communication) shows that for 1967, 304 SS and 316 SS accounted for approximately 60% of the total (1700 tons) usage of stainless steel pipe and tubing in the utility and chemical process and petroleum industries. The $2\frac{1}{4}$ Cr-1 Mo alloy steel is widely used for conventional power plant piping, boilers and superheaters, as well as for other purposes.

The selected materials are those which have been used most frequently in sodium systems and are favored for selection in LMFBR piping. In existing primary reactor coolant systems, 304 SS has been used in SRE, HNPF, EBR-II, EFAPP, and SEFOR; 316 SS has been used in LAMPRE-I and Rap-sodie. The only primary reactor coolant systems that did not use either of these two materials are EBR-I, SIR, and DFR (they used 347 SS) and BR-5, which used a titanium stabilized steel similar to 321 SS. Additional operating experience with these steels has been gained with sodium test facilities which have been operated in support of the development of these reactors. For the Aircraft Nuclear Propulsion program and other military projects, which required sodium system temperatures in excess of 1200°F, 316 SS was the favored material; a considerable amount of applicable data on 316 SS was generated during the development of these higher temperature systems.

*Tables and figures appear at the end of each section.

In the most recent studies of 1000-Mw(e) LMFBF's, the specified material for the primary systems is 304 SS. Similarly, 304 SS was recommended as a result of in-depth studies for the selection of structural materials for the FARET² and FFTF³ liquid metal systems.

In the C F Braun study for the SPTF,¹⁷⁴ 304H SS was recommended as the preferred material. In this study chromium-molybdenum ferritic steels, type 410 ferritic stainless steel, types 310, 321, and 347 austenitic stainless steels, and Incoloy 800 also were considered. Ferritic alloys were eliminated on the basis of low strength at 1200°F. Incoloy 800 was eliminated because some recent data indicated poor compatibility with sodium. Of the austenitic stainless steels, types 321 and 347 were eliminated because of structural instabilities (sigma formation), welding difficulties, and lack of sodium exposure data. Type 316 SS also was eliminated because of its greater tendency toward sigma formation than type 304 SS and because of its higher cost. Reference 174, as well as the earlier studies, References 2 and 3, present considerable detail on the preference of 304 SS and 316 SS over other candidate materials.

Both 304 SS and 316 SS are highly susceptible to stress-corrosion attack in aqueous media. Thus, their use in sodium-heated steam generators requires special precautions. To reduce the probability of failure due to stress corrosion, 2 $\frac{1}{4}$ Cr-1 Mo alloy steel has been used for the steam generators of HNPF, EBR-II, and EFAPP. The use of this alloy is also contemplated for future LMFBF steam generators. Because of possible economies, it would appear that the sodium piping material also might be 2 $\frac{1}{4}$ Cr-1 Mo at least for the cold leg.

Higher chromium ferritic alloys have been considered for secondary system cold leg piping, but because of welding difficulties and because of their higher cost than the 2 $\frac{1}{4}$ Cr-1 Mo steel they have been eliminated. Since the primary system with its intermediate heat exchanger will be 304 SS or 316 SS, the secondary system will be "bimetallic," if 2 $\frac{1}{4}$ Cr-1 Mo is used. In bimetallic secondary piping systems, the use of stabilized grades of 2 $\frac{1}{4}$ Cr-1 Mo steels could mitigate the carbon transfer problem, but such steels are not presently Code accepted in the United States and therefore cannot be used. Problems of mass transfer (particularly carbon transfer) and transition joints, which are special to such systems, are covered in this report.

1.4 OPERATING CONDITIONS

The range of steady-state operating conditions affecting the piping materials in an LMFBR system were presented in C F Braun & Co Technical Report 110 as Table 110-1, and is included as Table 1.1 here. Transient conditions and all abnormal conditions are to be defined at a later date.

TABLE 1.1 — RANGE OF STEADY-STATE OPERATING CONDITIONS

	Maximum Value	Minimum Value
Plant Life, yr	30	30
<u>Primary System</u>		
Sodium Temp, °F		
Hot leg	1200	900
Cold leg	970	600
Loop ΔT	300	230
Maximum Pressure		
Internal, psig	175	70
External, psig	20	0
Loop ΔP , psi	155	45
Piping pressure drop	—	—
Number of Loops	6	2
Total Flow, 10^6 lb/hr	127.4	16
Flow per Loop, 10^6 lb/hr	63.7	5.1
Flow per Loop, 10^3 gpm	149.2	12
Pipe Size, in.		
Hot leg	48	24
Cold leg	42	18
Cover Gas		
Kind	He or A	He or A
Purity, %	99.997	99.8975
Pressure, psi	30	3-in. H_2O
Sodium activity, r/hr	3.8×10^5	9.1×10^4
Radiation external to pipe, r/hr	3.8×10^5	—
Sodium Purity, ppm		
Oxygen	10	< 10 (Sat'n < 325°F)
Hydrogen	Sat'n < 325°F	0.002
Carbon	20	20

TABLE 1.1 — (CONTINUED)

	<u>Maximum Value</u>	<u>Minimum Value</u>
<u>Secondary System</u>		
Sodium Temp, °F		
Hot leg	1165	800
Cold leg	850	500
Secondary ΔT	315	250
Maximum Pressure		
Internal, psig	90	90
External, psig	0	0
Loop ΔP , psi	90	30
Number of Loops	4	2
Total Flow, 10^6 lb/hr	106.5	92
Flow per Loop, 10^6 lb/hr	46	5.0
Flow per Secondary Loop, 10^3 gpm	110	12
Pipe Size, in.		
Hot leg	44	16
Cold leg	42	16
Cover Gas		
Kind	He or A	He or A
Purity, %	99.997	99.8975
Pressure, psi	100	3-in. H_2O
Sodium activity, r/hr	1×10^{-3}	0.25×10^{-3}
Radiation external to pipe, mr/hr	<0.2	—
Sodium Purity, ppm		
Oxygen	10	—
Hydrogen	—	—
Carbon	—	—

Fig. 1.1 — Inter-Relationship Diagram — Task Area 3-8 — Coolant Transport and Storage Systems

2. COMPOSITION AND PROPERTIES OF SELECTED MATERIALS

2.1 INTRODUCTION

The information presented in this section is to provide a basis for the study of primary piping material properties in LMFBR environments. This basis, or reference point, includes the composition and physical and mechanical properties in air of the reference materials 304 SS, 316 SS, and $2\frac{1}{4}$ Cr-1 Mo. Physical properties, such as modulus of elasticity, coefficient of thermal expansion, thermal conductivity, and density are not expected to be affected by LMFBR environments, but mechanical properties may undergo significant changes when exposed to sodium as compared to air.

A high fast neutron flux environment also could have a significant effect on physical and mechanical properties. However, sodium piping is generally not expected to be exposed to such conditions. Exceptions may be certain sections of pipe in pool-type reactor systems. Extensive work on assessing the effects of radiation damage to the selected materials is being carried out under other elements of the LMFBR Program, e.g., in connection with the development of fuel cladding and core structural materials. These activities will not be covered in this report.

This section presents basic data on composition, metallurgical factors, physical properties, mechanical properties with emphasis on the effects of temperature for both short- and long-time air exposure, and basic weld properties for the selected materials. Of greatest importance are the effects of temperatures between 900 and 1200°F. Where available, data up to 1400°F are given.

2.2 COMPOSITION OF PRIMARY MATERIALS

Table 2.1 lists the various grades of the selected materials and their chemical composition. Both AISI and ASTM designations are given. Under the AISI designation, TP before the alloy number stands for tubular product grades. TP grades have the same compositional limits as regular AISI grades but are balanced to insure single-phase (100% austenite) structures for tube-piercing operations. This is also highly beneficial for elevated-temperature service.⁴

Acceptable ranges of alloying elements are given for each steel. For sodium piping, it may be desirable to specify alloy content within narrower ranges to assure control over certain metallurgical factors. Cobalt is limited to 0.10% in primary systems to minimize problems associated with radioactive mass transfer deposits.

The major difference between 304 SS and 316 SS is that the latter contains 2 to 3% Mo. This increases the strength of 316 SS over that of 304 SS above 900°F. Earlier data and the pre-1965 ASME Code, Section VIII, assigned significantly higher strength to 316 SS. However, more recent data and the 1965 code revision, have decreased the advantage of 316 SS over 304 SS. There is a sharp reduction in impact strength and ductility of 316 SS after long-time exposure at 1200°F, which is found to a lesser extent in 304 SS.

Three grades are shown for each of the two stainless steels (304 SS and 316 SS) in Table 2.1. The L suffix denotes low carbon (0.03% max), the H suffix denotes high carbon (0.04 to 0.10%), and no suffix denotes standard carbon (0.08% max). The low carbon content of the L grades limits carbide precipitation (sensitization) which may occur in the temperature range of 800 to 1400°F in unstabilized austenitic stainless steels with carbon above 0.04%. Thus, the L grades are less susceptible to intergranular attack by aqueous media. Elevated temperature mechanical properties are improved in the higher carbon level material. Therefore, the H grades should be specified in applications requiring maximum strength. The 1968 ASME Pressure Vessel Code¹¹⁴ limits the use of 304L to 800°F and 316L to 850°F. Since there is no lower carbon limit specified for the standard grades, it is recommended that these be avoided to obviate the possibility of obtaining low carbon material.

Since carbide precipitation does not appear to compromise corrosion resistance in sodium systems (except possibly during shutdowns and cleaning), it is probably desirable to take advantage of the better strength of the H grades for LMFBR piping. The material selected for SPTF piping is 304 H SS. SPTF is an isothermal, all stainless steel system in which there will be no tendency to transfer carbon because of uniform carbon activity throughout the system. In contrast, LMFBR primary piping is nonisothermal and the hot leg piping will tend to be decarburized [see Section 2.3.1(E)] due to the temperature effect on carbon activity.

Two 2 $\frac{1}{4}$ Cr-1 Mo steels are listed in Table 2.1. The A387-D normalized and tempered, has a minimum room-temperature tensile strength of 75,000 psi. In the annealed condition, this alloy has a minimum room-temperature tensile strength of 60,000 psi. The A335 Grade P22 can be obtained in the full-annealed, isothermal-annealed, or the normalized and tempered condition. In the an-

nealed condition, its minimum tensile strength at room temperature is 60,000 psi.

2.3 METALLURGICAL PROPERTIES OF PRIMARY MATERIALS

2.3.1 Austenitic Steels

304 SS and 316 SS have basically the same metallurgical properties. Both are austenitic stainless steels, and, in the annealed condition, have nearly a 100% austenitic or face-centered cubic structure at room temperature. Traces of ferrite, the body-centered cubic structure, may be present depending on compositional and fabrication factors. The subject of ferrite control, for welding and for sigma formation from ferrite, is discussed in Section 2.6.2(B). The amount of ferrite present depends on previous mechanical and thermal history and the exact composition of the heat. Elements that tend to promote the formation of ferrite are chromium, molybdenum, silicon, titanium, and columbium. Columbium and titanium are used in the stabilized stainless steels, i.e., 347 SS and 321 SS, which are not covered in this report. With the exception of molybdenum, the other elements will exist in both 304 and 316 SS in about the same range (see Section 2.2). The significant molybdenum addition in 316 SS (2 to 3% Mo) would make this alloy more ferritic than 304 SS, but is compensated by decreasing the Cr and increasing the Ni content. (See Table 2.1.) Other elements present in 304 SS and 316 SS, such as carbon, nitrogen, nickel, and manganese, behave as austenite promoters. The structure developed in a stainless steel at elevated temperature depends on time at temperature and the relative amounts of austenite and ferrite promoting elements.

Cold working of annealed austenitic stainless steel promotes formation of traces of both ferrite and martensite in proportion to the amount of cold work and basic composition. LMFBR piping will be in the annealed condition and is unlikely to undergo sufficient cold working during fit-up to transform to ferrite.

The principal metallurgical properties of the primary materials that will be discussed are: sigma-forming tendencies, carbide instability and precipitation, grain size, heat treatment, carburization, decarburization, nitriding, and denitriding. These are considered the most important properties influencing both the elevated-temperature and room-temperature properties after elevated-temperature exposure.

A. Sigma-Phase Formation

Sigma, an iron-chromium intermetallic embrittling phase, forms only in the chromium-containing stainless steels, both ferritic and austenitic, when

held in the temperature range of approximately 1000 to 1700 °F for extended periods of time.^{9,10} In 18-8 stainless steels, sigma forms most readily from ferrite^{11,12} but can also form directly from austenite.^{10,11,13} The rate of formation from austenite is much more sluggish.^{14,15} The time for sigma formation or nucleation from ferrite vs temperature is expressed graphically by a C-shaped curve.¹⁴ This curve has a point of inflection at a temperature where the time is at a minimum, usually in the order of minutes. Above and below this temperature, the time to form sigma increases. For example, for 304 SS and 316 SS the temperature for minimum time to form sigma is approximately 1100 and 1400°F, respectively,¹⁴ above and below which nucleation time becomes longer. The temperature for minimum time varies with alloy composition, the amount of cold work, and heat treatment. The time-vs-temperature curve for sigma formation from austenite is similar to that of ferrite.^{16,17} In annealed, fully austenitic 304 SS, sigma formation is retarded compared with cold-worked or ferrite containing material.

In the temperature range for LMFBR piping, sigma forms much more readily in 316 SS than in 304 SS.^{10,11} This is due to the 2 to 3% molybdenum present in 316 SS which acts as an elevated-temperature ferrite promoter¹⁸⁻²⁰ and sigma stabilizer.^{12,21} Greater amounts of ferrite result in a faster rate of sigma formation. Time to form metallographically detectable sigma in annealed and balanced 316 SS is on the order of 10,000 hr at 1200°F. A much longer time is necessary for sigma formation in 304 SS.^{16,17,22} Prior fabrication history influences sigma formation susceptibility and rate. Cold-rolled steels, hot-rolled steels, and annealed steels have decreasing order of susceptibility.^{11,23}

Both 304 SS and 316 SS respond to balancing of their compositions to reduce elevated-temperature ferrite and sigma.^{11,12,24} Balancing of 316 SS to compensate for its greater sigma-forming tendency leads to a recommendation for a maximum chromium equivalent (see Section 2.6.1) of 19% for 1200°F service.¹¹

A fine, evenly dispersed sigma, such as is formed from a cold-worked or a single-phase structure at 1100°F, enhances creep and rupture strength with little loss in initial toughness.¹¹ Continuous, agglomerated sigma at the grain boundaries, which develops around 1350°F or after very long aging, decreases creep and rupture strength and greatly reduces toughness.¹¹ Early stages of agglomerated discontinuous sigma do not have much effect on strength or toughness.

A fine grain structure increases the rate of sigma formation and greatly reduces toughness as compared to a coarse grain structure. A normalizing (~1850°F) heat treatment can reduce the effect of sigma on reducing toughness and possibly strength by redistributing and agglomerating existing ferrite. Sigma resulting from this ferrite during service will be more discontinuous and less harmful.^{12,24}

B. Carbide Instability

The principal carbide instability of possible concern is sensitization or carbide precipitation in the austenitic 304 and 316 stainless steels.

According to Binder,¹⁹⁰ when austenitic chromium-nickel steels are properly heat treated they exhibit a wide range of passivity in corrosive solutions. These steels resist corrosion most effectively when they are heated to temperatures within the range from 1050 to 1100°C (1920 to 2010°F) and cooled rapidly, because this heat treatment produces a homogeneous austenitic structure. Rapid cooling from the heat treating temperature is essential in order to keep the carbides in solution. Very small parts may be air-cooled but larger parts must be water-quenched. These steels, however, are subject to severe attack at the grain boundaries by even relatively mild corrosive mediums after exposure to temperatures in the range from 400 to 900°C (750 to 1650°F). This attack is known as "intergranular corrosion." The cause of intergranular attack has been the subject of considerable research. Some investigators believe that this kind of attack results from precipitation of chromium carbide and consequent depletion of chromium in the area adjacent to the grain boundaries. Decreasing carbon content decreases susceptibility of the steel to intergranular attack.

Besides carbon, the important factors are time, temperature, and corrodent. The data shown in Fig. 2.1 illustrate the influence of time and temperature on sensitivity to intergranular attack. It should be emphasized that the precipitation of carbides at the grain boundaries may take place very rapidly. For example, after welding, the metal adjacent to the weld is subject to intergranular attack, whereas the weld metal itself and the base metal outside the heat-affected zone are virtually free from attack. The intergranular weakness may be removed by annealing the steel properly or may be avoided by using stabilized compositions (types 321 and 347).

C. Grain Size

Grain size is affected by previous thermal treatment during processing. Finer grain size produces better room-temperature tensile, yield, fatigue, and impact strength.^{4,7} Table 2.2 shows the effect of grain size on tensile and yield strength. However, at elevated service temperatures, large grain size has a more beneficial effect on strength as measured by creep rate and rupture strength.⁴⁻⁷ For high-temperature stainless steel piping, an ASTM grain size of 5 to 7 is generally acceptable, but further study of possible benefits from specifying larger grain sizes for LMFBR sodium piping is indicated. The desired grain size is obtained by specifying thermal treatment and fabrication procedures during processing of stainless steels by the mill and pipe fabricator.

The effect of grain size may be obscured by other factors affecting elevated temperature properties such as prior heat treatment, sigma, carbide precipitation, and minor variations in chemical composition.⁵ Grain growth which occurs at elevated temperatures depends on the degree of prior work-hardening as well.^{5,9}

D. Heat Treatment

The austenitic stainless steels (e.g., 304 SS, 316 SS, and their modifications) are normally procured in the annealed state when intended for LMFBR service. During processing from ingot to pipe, hot-working and periodic intermediate annealing treatments are employed. In general, the temperature-time parameters of the final heat treatment are adjusted to produce the desired grain size and microstructure.

The manipulations required during fitup and welding in field erection operations produce residual stresses and microstructural changes, but there are no Code requirements for a stress relief heat treatment after fabrication.

Optional heat treatments are given in the Codes to obtain special corrosion properties and/or dimensional and structural stability of assemblies. They include full anneal, high-temperature stabilization, and low-temperature stress relief.

Heat-treating temperature can have a profound effect on rupture strength and ductility, particularly when carbon content is on the high side of the permissible range. Fig. 2.2 demonstrates this interaction which was investigated by Goodell et al.¹¹⁷

E. Carburization and Decarburization

The high-chromium content of austenitic stainless steels renders them very susceptible to carburization because the chromium carbides are very stable compounds. The result of substantial carburization is severe embrittlement of these steels and possible impairment of stainlessness. In addition, the steels may be decarburized under conditions of low-carbon activity which may be produced by large temperature differences in a sodium system.

Carbon may transfer to or from piping materials carrying sodium at LMFBR temperatures. Background for such transfer is provided in Section 3.2.2(B). The effects of the resulting carburization and/or decarburization on the mechanical properties of 304 SS and 316 SS are covered in Section 3.3.

Although carburization is of great concern, there does not appear to be a widely accepted limitation for the maximum carbon content of austenitic stain-

less steel. In particular, the effects of long-time aging on properties have not been measured for various carbon contents. It is known that aging at temperatures in excess of 1000°F produces carbide precipitation with profound effects on ductility and toughness.

Carbon movement within the steels is determined by the diffusivity of the carbon in the specific environment. Environmental factors include temperature, stress fields, and phase relationships.

Smith³³ has reported diffusivities of carbon in austenite composed of iron and nickel. These alloys may simulate the condition in austenitic stainless steels after full precipitation of chromium. Fig. 2.3 presents an extrapolation of Smith's data which were obtained at higher temperatures for a 23% nickel and 0.1% carbon composition and which were determined by a decarburization method. These values are two orders of magnitude larger than those reported by AI³² and MSA.^{100,102} Although the pseudo-diffusivities computed at AI and MSA are too low, their activation energies are in good agreement with those cited by Smith.

Campbell and Tyzack³⁴ proposed a model for the carburization of stainless steels in a medium of unit carbon activity. They assumed precipitation of chromium and used auxiliary data to estimate carbon activities at various stages of the process. With this model they recalculated the diffusivity obtained at AI, but their corrected value is still in error by a substantial amount.

Kolodney³⁵ employed a carburization rate parameter to correlate the results of various investigators as well as preliminary carburization rate measurements made at United Nuclear Corporation using a carbon meter. He proposed a simplified theoretical equation for the rate of formation of a carburized case and showed that it corresponds fairly well with measured values.

F. Nitriding and Denitriding

In the past, nitrogen was not an intentional addition to austenitic stainless steels but was acquired principally by reaction between the chromium in the melt and nitrogen in the air. Its effects are generally beneficial in maintaining high-temperature creep and rupture strength²⁶ and low-temperature toughness. It is a powerful austenite stabilizer, and accordingly, ferrite and sigma suppressor.^{11,13,18} Avery,¹⁸ however, states that above 0.20% nitrogen, unexpectedly higher creep rates are produced in 26 Cr-12 Ni stainless steel and recommends 0.20% as a maximum limitation.

Recently, nitrogen-bearing stainless steels have been approved¹⁴³ for use in welded construction under the rules of Sections I, III, and VIII of the ASME Boiler and Pressure Vessel Code. Compositions, minimum mechanical

properties, and design stresses are presented in the Code Case 1423-1 which is included as an Appendix.

Goodell et al.¹¹⁷ have investigated the effect of nitrogen on the creep-rupture properties of 304 SS. They found that increasing nitrogen content resulted in greatly increased 1200°F-1000 hr rupture strength (see Fig. 2.4) for all carbon contents between 0.01 and 0.14% and that the increase in strength from a given nitrogen addition was greater than that due to the same amount of carbon. Ductility suffered less from nitrogen additions than from carbon additions. In contrast with behavior with carbon, the strengthening effect of nitrogen appeared to be nearly independent of heat treatment. The extrapolated 100,000-hr rupture strength for 304 SS with nitrogen and carbon additions correlates with the parameter: $(\%C) + 1.25 (\%N)$, see Fig. 2.5.

Due to the generally beneficial effects of normal levels of nitrogen on the elevated-temperature properties of 304 and 316 SS mentioned above, loss of nitrogen is possibly more undesirable than moderate interstitial increases.

Nitrogen may be transferred to or from piping materials by sodium at LMFBF temperatures. This is covered in Section 3.2.2(C).

2.3.2 Ferritic Steels ($2\frac{1}{4}$ Cr-1 Mo)

These steels exist in the ferritic or body-centered cubic crystalline form at room temperature. They obtain their superior room temperature strength properties vs plain carbon steel by solid solution strengthening and increased hardenability from the additions of chromium and molybdenum at the same carbon levels of low carbon steels. The elevated temperature properties are also enhanced by the alloy additions making these alloys suitable for elevated temperature service.^{5,37}

A. Alloying Effect on Properties

The addition of molybdenum has the greatest effect on increasing elevated-temperature strength, particularly in the range of 600 to 1100°F.^{5,37,38} Chromium, although of less importance, provides some additional high-temperature strengthening effects up to around $2\frac{1}{2}\%$, compared to low carbon steel with molybdenum alone. Chromium acts to improve oxidation resistance, to retard spheroidization, and to prevent graphitization in $2\frac{1}{4}$ Cr-1 Mo steel.

The addition of chromium and molybdenum produces an air-hardening alloy resulting in a hardened fusion and heat-affected zone after welding (or any local heating above about 1600°F), and accordingly, requiring stress relief to remove the local embrittling effect which acts as a metallurgical notch.^{5,37,39} This air-hardening consideration, as well as the need for room-temperature ductility and good weld properties, limits the carbon content to 0.15% maximum.

B. Long-Term Aging Effects

The $2\frac{1}{4}$ Cr-1 Mo slowly spheroidizes during long temperature service above 900°F. This results in a decrease in tensile properties with a proportionate increase in ductility.⁵ The impact properties increase as a result. The data showing the effect on mechanical properties are presented in Sections 2.5.1 and 2.5.2.

C. Decarburization

The mechanical properties of $2\frac{1}{4}$ Cr-1 Mo steel are affected in a significant way by changes in carbon content, with decarburization having a much more serious effect than carburization. Carbon transfer resulting from contact with sodium under LMFBR conditions is covered in Section 3.2.2(B). The effects of decarburization on selected mechanical properties in $2\frac{1}{4}$ Cr-1 Mo are presented in Sections 3.3.1(F) and 3.3.2(E).

As indicated in Section 2.3.1(E), the movement of carbon within the steel is determined by the diffusivity of the carbon in the specific environment. The diffusion coefficient for carbon in ferrite (alpha iron) as determined by Homan¹¹² is shown in Fig. 2.6.

Even during routine heat-treating operations, surface decarburization often results. Koch and Kovach¹¹⁸ found this when $2\frac{1}{4}$ Cr-1 Mo piping used in a model sodium-heated steam generator was heat treated in accordance with ASME Code requirements. Preferential oxidation of chromium and silicon also was found.

The driving force for carbon migration is the carbon activity gradient. Carbon activity in a given steel is determined by how tightly it is bound. The high stability of chromium carbide results in low-carbon activity in chromium-bearing steels. Fig. 2.7 shows the effect of chromium content on the activity of carbon in the Fe-Cr-C system as reported by Herberg.¹⁴⁴

D. Heat Treatment

The alloy content of the $2\frac{1}{4}$ Cr-1 Mo steels is sufficient to produce air-hardening characteristics on cooling from the temperatures involved in hot forming and welding operations. The hot-forming temperatures (1750-1850°F) parallel the standard normalizing heat treatment range and thus require stress-relieving or tempering treatments following such operations to reduce as-cooled hardness levels and relieve residual forming stresses. Tempering is performed in the 1250 to 1425°F ranges, while stress relieving temperatures of 1300 to 1400°F are usually employed.

Normal practice and code⁴⁰ specifications for welding to these steels require preheating of the facing members to prevent excessive hardening and cracking of the fusion and heat-affected zones of the joint. A 300°F minimum preheat is permitted with use of low hydrogen coatings on metal arc electrodes. Alternate electrode coatings and welding processes necessitate an increase in preheating temperature to 400°F. Similarly, post-welding heat treatments are required prior to service operation. The stress-relief treatment specified by the several codes and the AWS Handbook involves heating to 1250 to 1425°F for approximately 1 hr/in. of material thickness followed by slow cooling to ambient temperature. A summary of heat-treatment cycles and transformation temperatures are noted in Tables 2.3 and 2.4.

For the case of dissimilar material welds involving 2 $\frac{1}{4}$ Cr-1 Mo and austenitic stainless steels, the preheat and postheat treatments normally employed for the ferritic alloy are specified.

2.4 PHYSICAL PROPERTIES OF PRIMARY MATERIALS

2.4.1 Elastic Constants

Data on modulus of elasticity, shear modulus, and Poisson's ratio are presented vs temperature in Tables 2.5, 2.6, and 2.7, respectively. The data for modulus of elasticity are taken from USAS B31.7.⁴² They are identical to those listed in Reference 47. Reference 47 also is the source for the data in Tables 2.6 and 2.7. The modulus of elasticity was determined by the static bend test method and the shear modulus by the static torsion test method.

2.4.2 Density

The density at room temperature of the austenitic stainless steels in the annealed condition is 0.29 lb/in.³ and that of 2 $\frac{1}{4}$ Cr-1 Mo, 0.28 lb/in.³ The variations of density with temperature calculated from thermal linear expansion data and the measured density at 32°F for 304 SS, 316 SS, and 2 $\frac{1}{4}$ Cr-1 Mo are shown in Table 2.8.

2.4.3 Thermal Properties

Table 2.9 summarizes mean coefficient of thermal expansion data for the primary materials over various temperature ranges to 1200°F. Table 2.10 presents instantaneous values.

Figs. 2.8 and 2.9 are plots of the instantaneous and mean coefficients of thermal expansion of the same material vs temperature. Fig. 2.10 and Table 2.11 show the effect of temperature on the thermal conductivity of the primary materials.

Specific heats of annealed steels at various temperatures are given in Table 2.12.

The calculated thermal diffusivities of the primary materials are presented in Table 2.13.

2.5 MECHANICAL PROPERTIES OF PRIMARY MATERIALS

2.5.1 Tensile Properties

The tensile properties from room temperature to elevated temperatures are given for short times at the higher temperatures. The stress-strain relationship for the primary materials departs from proportionality at relatively low values of stress and there is no sharp yield strength. The 0.2% offset yield strength is used. Ranges of tensile and yield strength and percent elongation vs temperature curves for the primary materials from room temperature to elevated temperatures are shown in Figs. 2.11, 2.12, and 2.13 for 304 SS, 316 SS, and $2\frac{1}{4}$ Cr-1 Mo.

Recently, Smith¹⁴⁵ evaluated existing high-temperature data for the austenitic stainless steels to develop trend curves for yield and tensile strength. These are presented in Figs. 2.14 and 2.15.

Typical ductility data, expressed as elongation percent in 2-in. gauge length after tensile testing vs temperature, are presented in Table 2.14.

Campbell¹¹⁵ in a review of recent developments concerning mechanical properties of metals reported on GE work on stress-strain behavior of stainless steels. In work on the effect of strain rate on tensile properties, it was shown that at elevated temperatures increased strain rate adversely affected the tensile strength, the reduction in area, and the ductility. Table 2.18 and Fig. 2.16 show this effect.

Room-Temperature Tensile and Elevated-Temperature Rupture Properties after Elevated-Temperature Aging

These data are presented in Tables 2.15, 2.16, and 2.17 for 304 SS, 316 SS, and $2\frac{1}{4}$ Cr-1 Mo, respectively. They show the following effects due to aging:

- 304 SS

A slight gain followed by a slight loss in room-temperature tensile properties and a slight loss in ductility with a small increase in rupture strength.

- 316 SS An increase in tensile properties, a decrease in ductility, and a corresponding increase in rupture strength.
- $2\frac{1}{4}$ Cr-1 Mo A decrease in all properties except ductility, which increases.

These observations for 304 SS and 316 SS can be explained by elevated temperature aging effects including carbide precipitation and sigma formation during the 10,000-hr aging at 1200°F. The loss in impact strength and ductility in 316 SS is more attributable to the effects of sigma formation. In 304 SS it is probable that carbide precipitation is a major contributor to diminished ductility and toughness in the time periods involved. The decrease in strength for the $2\frac{1}{4}$ Cr-1 Mo can be accounted for by the spheroidization of the initial pearlite during holding at 900 to 1200°F for long times.

2.5.2 Impact Properties

The impact properties for 304 SS and 316 SS materials at room temperature and at several elevated temperatures are shown in Figs. 2.17 and 2.18. Values were obtained by the Charpy test method on keyhole notched specimens. Data were obtained with bar stock in the annealed conditions. Impact strength will be lower for work-hardened conditions. The data for one hour at temperature show a generally downward trend in impact values for both steels with increasing temperature. Table 2.19 gives impact data for $2\frac{1}{4}$ Cr-1 Mo.

Impact Properties after Elevated-Temperature Aging

A considerable amount of data is available in Reference 22 on the effects of elevated temperature aging on reducing the toughness of 304 SS and particularly 316 SS at room temperature. Figs. 2.19 and 2.20 illustrate this loss in room-temperature toughness after high-temperature exposure (900 to 1500°F) for 304 SS and 316 SS, respectively. Tables 2.15 through 2.17, previously presented, show similar impact data on the effects of elevated-temperature aging on 304 SS, 316 SS, and $2\frac{1}{4}$ Cr-1 Mo for times up to 10,000 hr. The trends are the same for 304 and 316 SS. The aging data in Table 2.17 on $2\frac{1}{4}$ Cr-1 Mo show an increase in toughness at room temperature, after elevated-temperature exposure, in contrast to the decreases shown for the stainless steels.

Data obtained on 316 SS after 60,000 hr aging and extrapolated to 100,000 hr are presented in Table 2.20.^{22, 11} These data indicate the marked loss in impact strength in going from 10,000-hr data to 100,000-hr extrapolations at temperatures of 1150 to 1250°F. This loss in impact strength must be carefully considered for LMFBR applications where 30-yr life is planned. Specific compositions of 316 SS within the specification range might be required.

2.5.3 Fatigue Properties

In design against fatigue failures we are concerned with: (1) high-cycle fatigue, in which the nominal cyclic stresses are below the yield strength of the material and the number of cycles to failure is of the order of 10^6 ; and (2) low-cycle fatigue in which there are conditions of cyclic plastic strain and the number of cycles to failure is of the order of 10^4 or less. In addition, at elevated temperatures the material behavior becomes much more complicated due to creep and diffusion controlled reactions.

In the creep range, it has been shown that frequency has a strong effect on fatigue properties.^{146,147} This is the same thing as saying that the strain rate is important, and related to this the hold-time in the cycle is important, with fatigue life decreasing as hold-time increases.

Another problem, applicable over the whole temperature range, is cumulative fatigue damage, where cyclic strains of greatly different magnitudes are experienced, and in varying and probably random order of application. The order of application, as well as the difference in magnitude, may have a significant effect on the overall fatigue life.

The general procedure for obtaining experimental fatigue data is to control the strain between fixed limits, and if desired, to measure the stress, while cycling the specimen until failure is determined. Such tests may be under conditions of controlled diametral strain, longitudinal strain, bending strain, or torsional strain. The results can be represented in terms of total strain range vs cycles to failure, or, if desired, of plastic strain range vs cycles to failure. The former is most often used for direct design procedures and has been applied in the fatigue design curves of the ASME Code, Section III.

There is a limited amount of stress-controlled data, representing conditions which might occur in the case of mechanical loads due to vibration. However, since in most cases stresses causing fatigue are due to thermal gradients or local discontinuities, they are self-limiting, and therefore, strain controlled.

Figs. 2.21 and 2.22 show curves of strain range/2 vs cycles to failure for 304 SS and 2 $\frac{1}{4}$ Cr-1 Mo steels. Data were obtained from strain-controlled tests where specimens were cycled under conditions of constant longitudinal strain until failure. The equivalent alternating stress, S_a , was determined from the maximum strain measurements by applying the formula:

$$S_a = 1/2 E_t \epsilon_R$$

where S_a = alternating stress in psi

ϵ_R = total strain range at test temperature

E_t = modulus of elasticity at test temperature.

Corresponding curves of alternating stress, S_a , vs cycles to failure are given in Figs. 2.23 and 2.24.

Fig. 2.25 shows a comparison of fatigue failure curves for types 304, 348, and 316 stainless steels. All data fit closely to the same curve, indicating that one fatigue curve can apply to both types 304 and 316. Figs. 2.26 and 2.27 show the effects of several temperatures and strain rates on the fatigue life of type 304 and 316 materials. Fig. 2.28 shows a limited amount of stress-controlled fatigue data for type 304 stainless steel. It can be seen that, in general, the fatigue life is less when a constant stress (or load) is maintained than for a constant strain.

Endurance limits for the materials discussed above are given in Table 2.21. These values represent the asymptotic or flat portions of the fatigue curves at room temperature, and were estimated from the fatigue curves for each material.

Campbell¹¹⁵ has reviewed GE work¹⁷¹ in which tensile data have been applied to predict low cycle fatigue behavior of 304 SS at 1200°F and at a strain rate of 4×10^{-3} per second using the following equation to obtain the $\Delta \epsilon_t$ vs N_f relationship:

$$\Delta \epsilon_t = \frac{\frac{3.5 \sigma_u}{E} N_f^{-0.12} \pm \sqrt{\left(\frac{3.5 \sigma_u}{E} N_f^{-0.12}\right)^2 + \frac{2 \epsilon_f^2}{N_f}}}{2}$$

where $\Delta \epsilon_t$ = the total strain range

σ_u = the tensile strength

N_f = the number of fatigue cycles to fracture

E = the elastic modulus

ϵ_f = the tensile ductility, $\ln (100/100-RA)$, where RA is the percent reduction in area.

The points in the curve of Fig. 2.29 represent experimental fatigue data obtained at the same strain rate.

2.5.4 Creep and Stress to Rupture

At elevated temperatures, where creep is involved, the deformation depends on both the applied stress and the time of its application. It is customary to limit design stresses to a value that will cause a specified strain or deforma-

tion in a given time, or a stress that will cause rupture in a given time. The most generally used values are the stress to cause rupture in 100,000 hr and the stress to produce a secondary creep rate of 0.01%/1000 hr. In his evaluation of existing creep and rupture data for austenitic stainless steels, Smith¹⁴⁵ has used visual extrapolation and regression analysis (with both time and stress as the independent variable) to arrive at average and minimum values of these stresses. These are shown in Tables 2.22 and 2.23 and in Fig. 2.30. For LMFBFR, 200,000-hour rupture data and creep data are required. Tables 2.24, 2.25, and 2.26 provide such data based on Smith's evaluation. Data for $2\frac{1}{4}$ Cr-1 Mo are also shown in Tables 2.24 and 2.26 and in Figs. 2.31 and 2.32.

Smith¹²² also has observed linearity between log applied stress and log average creep rate (derived as the quotient of elongation-at-rupture by time-to-rupture) and has used this observation to extrapolate rupture ductility. Results of this approach are plotted in Fig. 2.33.

Another approach to presenting tensile and creep data is that proposed by Manjoine.¹²¹ At a given temperature the total strain resulting from a stress applied for a given period of time is plotted against the stress. When the data for each stress and time period are plotted, curves are drawn through the strains obtained for equal time resulting in "isochronous stress-strain curves" (see Fig. 2.34). The curve for "zero time" is the conventional stress-strain curve up to the maximum strain of interest. From such curves, the creep strength for a given strain and time can be read directly.

Fig. 2.34 illustrates the complete manner in which short-time tensile, creep, and rupture data can be presented. The curve for $t = 0$ is the initial portion of the short-time stress-strain up to 7% total strain. The ultimate strength is given by the symbol U on the right edge. The curve e_0 , is the strain intercept of the creep curve at zero time. Isochronous curves are shown for time periods of 100, 1000, 10,000, and 100,000 hr. These curves "flatten out" at the rupture strength for these times and this rupture stress is indicated by the symbol R at the right edge above the corresponding curve. The creep strength for a given strain and time can be read directly from the curves.

The effects of carbon and nitrogen individually and in combination on the 1000-hr rupture strength were discussed above. Figs. 2.2, 2.4, and 2.5 present the effects of these elements.

2.6 WELDING PROPERTIES OF PRIMARY MATERIALS

All the piping joints for the LMFBR loop systems are to be fusion-welded. Accordingly, the weldability of the primary materials and the effects of welding on the metallurgical and mechanical properties of the joint will be of primary concern. Compositional changes of the weld metal compared to the base material receive consideration, including ferrite content of stainless steel welds and the case of dissimilar welds. This section does not encompass technique, joint configuration, or welding processes.

2.6.1 Austenitic Steels – 304 SS and 316 SS

A. Composition of Weld Filler Metal

For 304 SS piping, 308 SS (20 Cr-11 Ni) filler rod is commonly used and is most applicable to LMFBR piping because its chemistry minimizes cracking and limits sigma-forming tendencies in the weld.

For 316 SS piping, a weld filler metal composition (17 Cr-12 Ni-2.5 Mo) essentially the same as base material, or a 16 Cr-8 Ni-2 Mo modification is used. The subject of delta ferrite control in welds is discussed below for both 304 and 316 SS.

It must be noted that RDT Specifications F6-IT(S902), Welding Standard, does not permit the use of the manual shielded metal-arc process on the first, second, and final layers of welds on components enclosing liquid metals.

B. Metallurgical Effects

The heat energy input essential for fusion welding produces microstructural changes in the heat-affected zones adjoining the joint. Weld metal and heat-affected zones will be subject to carbide precipitation (sensitization) by the heat input for subsequent weld layers in multipass weld joints. The degree of sensitization and its resultant effect on the corrosion behavior of the weld and adjacent material is a direct function of the time spent in the temperature range and the temperature level attained, as is illustrated in Fig. 2.35.

Inspection of this figure indicates that for short-term exposures the peak effect occurs at 1100 to 1300°F. At lower temperatures, longer time exposures are required to reach peak sensitization. Exposure times beyond those indicated to develop peak sensitization produce a reduction in corrosion effects. For LMFBR cold leg temperatures of 850°F, more than 10,000 hr would be required to achieve such a reduction. Experience in steam systems has shown that the modifications in grain structure and sensitization in the welds and adjacent zones can be expected to have limited influence on the performance of the austenitic stainless steels in LMFBR service.

The use of adjusted filler metal compositions, to develop crack-free welds and to limit subsequent transformation of ferrite to sigma phase during service, requires further definition.

Control of delta ferrite content in austenitic stainless steel weldments is essential to the achievement of weld metal resistant to cracking on cooling from the molten state. Experience has shown that a minimum of 3 to 5% delta ferrite is required in the filler metal to produce good weldability. Under LMFBR service conditions of 1100 to 1200°F, the ferrite is known to transform progressively to sigma phase with resultant reduction in toughness and ductility. A maximum of 10% ferrite has been established by Malone¹² as an effective upper limit within which acceptable material ductility can be maintained even after extended service time.

Actual control of the amount of delta ferrite developed in a particular weld filler composition is achieved by adjustment of the relative quantity of ferrite and austenite stabilizing elements. For gas metal arc electrodes the various constituents may be incorporated in either the core wire or coating. For gas tungsten arc welding it is necessary to adjust filler wire composition to effect the desired ferrite content. Use of the Schaeffler diagram (Fig. 2.36) permits prediction of delta ferrite content of as-deposited weld metal from the chemistry of the weld rod. Delta ferrite content can be determined with acceptable accuracy by use of a Magne-Gage instrument.

Definitive determination of delta ferrite in production welds necessitates evaluation of the chemistry of piping materials and projected dilution of the weld filler metal with base metal during joining.

Code requirements as outlined in USAS B31.7, Nuclear Power Piping, and ASME Pressure Vessel Code, Sections III and VIII, neither prohibit nor require pre- or post-heat treating of welds joining the austenitic stainless steels. Industry practice has been to warm the components to 60°F minimum, if ambient temperatures are lower, prior to the start of welding and perform no post-heating wherever the possibility of corrosion in service is a design consideration. The Welding Handbook¹⁴⁹ recommends various post-heating treatments depending on environmental¹⁵⁰ and design considerations in service. The high thermal expansion coefficients of the austenitic stainless steel alloys produce significant residual stresses in weldments, particularly under conditions of high restraint and heavy wall thickness. Under service conditions, where corrosion is not a major concern, the application of one of the several post-weld thermal treatments outlined by the Welding Handbook¹⁵⁰ should be considered.

C. Mechanical Properties

• Tensile and Creep Properties vs Base Metal

Type 304 SS and 316 SS weldments possess tensile properties (ultimate strength, yield strength, ductility) in the same range as the corresponding base metals.^{45,180} The same holds true for stress-rupture properties for times up to 100-hr duration. At rupture times exceeding 100 hr and in creep the weld materials show strength near and below the minimum levels determined for wrought alloys.^{45,180,181} This loss of comparable rupture and creep strength becomes more pronounced at 1200°F and higher temperatures. It should be noted that existing piping and pressure vessel codes do not recognize this difference between weld metal and base alloy properties.

• Toughness (Charpy Keyhole Notch)

Type 316 SS weldments show a very significant decrease in toughness at room temperature in the weld metal, and to a lesser extent in the heat-affected area vs the base metal.²² Solution annealing followed by air cooling increases the toughness in the weld area to a significant degree, but the value remains only one-half that of the base metal. The effect of long-term aging (10,000 hr) shows an even greater relative decrease in toughness in the weld metal vs the base material, which also decreased.

2.6.2 Ferritic Steels – 2 $\frac{1}{4}$ Cr-1 Mo

A. Composition of Weld Metal

Weld metal composition is essentially the same as the base metal. A low hydrogen coated electrode is used for the metallic arc process.

B. Metallurgical Effects

Since 2 $\frac{1}{4}$ Cr-1 Mo is an air-hardening alloy, the weld metal and heat-affected zones will be harder after weld cooling than the base material. A pre-heat and interpass temperature of 300°F (minimum) is necessary for successful welding to limit the level of hardening developed by welding heat input. A 1300 to 1425°F stress relief at 1 hr/in. of thickness will be necessary according to Code requirements on all piping over 4 in. in diameter to eliminate this metallurgical notch which could lead to failure by cracking. Stress relief is not required by the Codes for smaller diameter piping with less than 1/2-in. wall

thickness where proper welding control is utilized and the carbon content of the material has been limited to 0.15% maximum.¹⁴⁸

C. Mechanical Properties

A study⁴⁵ made on welded specimens of $2\frac{1}{4}$ Cr-1 Mo showed virtually no difference in strength in welded specimens, compared to base metal strength ranges. The creep properties, the short-term and stress rupture properties were essentially unchanged. Ductility of welded specimens was somewhat lower. Most of the tensile ruptures on the welded specimens occurred in the base material on defect-free welds. No toughness or fatigue data were available.

2.6.3 Dissimilar Joints (304 SS or 316 SS to $2\frac{1}{4}$ Cr-1 Mo)

There are several possible solutions to the problem of welding austenitic stainless steel to $2\frac{1}{4}$ Cr-1 Mo. Each has its advantages and limitations.

A Ni-Cr-Fe composition such as Inconel 82 or 182* is recommended as a weld filler alloy, but requires careful control during welding and is more subject to nickel solution attack and mass transfer by sodium than the chromium-nickel stainless steels. An alternate filler metal is E-309 stainless steel.¹⁸² Dilution of chromium and nickel of the E-309 with the $2\frac{1}{4}$ Cr-1 Mo base metal form a series of air-hardenable alloy compositions at the edge of the weld fusion zone. Pre- and post-heating are essential to prevent cracking. A variation to the above is to overlay the $2\frac{1}{4}$ Cr-1 Mo side with E-309 followed by welding with E-308 or E-309 filler. The latter procedure minimizes the dilution of austenitic weld metal with the ferritic base alloy.

One of the major problems associated with welds joining austenitic stainless steels to ferritic steels is the significant difference in thermal expansion coefficients as shown in Table 2.27. Under thermal cycling in service this difference can produce stresses sufficient to cause failure.¹⁷⁵ Use of austenitic stainless steel weld compositions, such as 309 SS (25 Cr-12 Ni) does not modify this difference and places the full thermal stress at the hardenable austenitic-to-ferritic interface of the weld joint. Use of a filler weld metal composition such as Ni-Cr-Fe (see Table 2.27), having a thermal expansion coefficient between those of the ferritic and austenitic alloys, produces two important effects. The level of thermal stresses is reduced, and the stresses are at least partly placed at the ductile austenitic-to-Inconel weld interface.¹⁷⁶

*Huntington Alloys Division, International Nickel Company.

The Welding Handbook¹⁷⁷ and other reports^{175, 176, 178, 179} recommend use of such modified filler metal compositions. However, extensive successful industry experience exists with use of E-309 SS filler alloy for dissimilar welds involving austenitic stainless and 2 $\frac{1}{4}$ Cr-1 Mo steels, particularly at operating temperatures below 750°F.

A. Composition of Weld Metal

The most satisfactory weld filler alloys are of the Ni-Cr-Fe type such as "Inconel 82" for GTA welds and "Inconel 182" for metal arc welding. Both alloys are basically Ni: ~70%, Cr: ~15%, Fe: ~8%, modified with additions of manganese and columbium. These alloys in LMFBR piping above 900°F could be subject to preferential sodium solution attack and mass transfer of nickel.

An alternate filler metal alloy to Ni-Cr-Fe compositions is E-309, which contains principally 22.00 to 25.00 Cr, 12.00 to 14.00 Ni, and balance Fe. It would be used by buttering the ferritic alloy side with 309 SS and completing the joint with either 309 SS or 308 SS filler material. However, the 300 series stainless alloys are not generally recommended for joints in service above 900°F where more severe thermal stress cycling might occur because of the effects of differential expansion. In addition, carbon migration across the ferritic/austenitic stainless steel fusion line becomes more pronounced above 900°F as diffusion rates increase. These problems are not as severe with Ni-Cr-Fe type filler metals.

B. Metallurgy

Ni-Cr-Fe type filler metals have none of the special metallurgical problems associated with the alloys discussed above for 2 $\frac{1}{4}$ Cr-1 Mo steel, 304 SS, or 316 SS, i.e., air hardening, ferrite control, and sigma-forming tendencies. However, the 2 $\frac{1}{4}$ Cr-1 Mo side of the weld joint will air harden, and pre-heat and post-heat treatments must be used.

A 309 filler will have basically about the same metallurgical properties as 308 SS and 316 SS fillers described above. The tendency to form sigma will be greater than 308 SS filler and perhaps greater than 316 SS filler.

C. Physical Properties

Differential thermal expansion and thermal conductivity are the physical properties of major importance for the above dissimilar welds. The total effects of differential expansion across the joint will become considerable starting at 900°F. The difference in thermal expansion is particularly large be-

tween $2\frac{1}{4}$ Cr-1 Mo and all the stainless steels discussed (309 SS, 304 SS, and 316 SS). Inconel 82/182 has a coefficient between those of $2\frac{1}{4}$ Cr-1 Mo and 304 SS and its use helps to reduce the effects of the differential thermal expansions. The effects of these differential expansions will be to accelerate thermal fatigue failure at the weld joint in cyclic thermal service.³⁹

D. Mechanical Properties

Generally, it can be expected that the joint, if sound, will be at least as strong, ductile, and tough as the poorer of the two base materials. The filler metal (i.e., Inconel or 309 SS) will have mechanical properties either lying between the two dissimilar base materials or exceeding the weaker. However, failure during a specific tensile or creep test is likely on the fusion line, or in the heat-affected zone, particularly on the ferritic alloy side.^{39, 45, 46} The fatigue behavior of these joints is unknown.

Limited data on tensile properties of a $2\frac{1}{4}$ Cr-1 Mo to 316 SS welded joint are presented in Table 2.28. The $2\frac{1}{4}$ Cr-1 Mo side of the joint was first buttered with 309 SS before making the final joint with 309 SS filler rod.⁴⁶ These data show that the rupture occurred in the 316 SS base metal, which is indicative of a satisfactory weld.

TABLE 2.1 — CHEMICAL COMPOSITION OF SELECTED ALLOYS

Alloy	Designation				Chemical Composition, %							
	AISI	ASTM (Welded Pipe)	ASTM (Seamless Pipe)	ASTM Plate	C	Mn	Si (max)	Cr	Ni	P(max)	S (max)	Mo
304L	TP 304L	—	—	A 240	0.03 max	2.00	1.00	18.00 to 20.00	8.00 to 12.00	0.045	0.030	—
		A 312	A 312	—	0.035 max	2.00	0.75	18.0 to 20.0	8.00 to 13.0	0.040	0.030	—
304	TP 304	—	A 376	—	0.08 max	2.00	0.75	18.0 to 20.0	8.0 to 11.0	0.030	0.030	—
	—	A 312	A 312	—	0.08 max	2.00	0.75	18.0 to 20.0	8.00 to 11.0	0.040	0.030	—
	—	—	—	A 240	0.08 max	2.00	1.00	18.00 to 20.00	8.00 to 12.0	0.045	0.030	—
304H	TP 304H	A 312	A 312	—	0.04/0.10	2.00	0.75	18.0 to 20.0	8.00 to 11.0	0.040	0.030	—
	—	—	A 376	—	0.08 max	2.00	0.75	18.0 to 20.0	8.0 to 11.0	0.030	0.030	—
316L	TP 316L	—	—	A 240	0.03 max	2.00	1.00	16.00 to 18.00	10.00 to 14.00	0.045	0.030	2.00 to 3.00
	—	A 312	A 312	—	0.035 max	2.00	0.75	16.0 to 18.0	10.0 to 15.0	0.040	0.030	2.00 to 3.00
316	TP 316	—	—	A 240	0.08 max	2.00	1.00	16.00 to 18.00	10.00 to 14.00	0.045	0.030	2.00 to 3.00
	—	A 312	A 312	—	0.08 max	2.00	0.75	16.0 to 18.0	11.0 to 14.0	0.040	0.030	2.0 to 3.0
	—	—	A 376	—	0.08 max	2.00	0.75	16.0 to 18.0	11.0 to 14.0	0.030	0.030	2.0 to 3.0
316H	TP 316H	A 312	A 312	—	0.04 to 0.10	2.00	0.75	16.0 to 18.0	11.0 to 14.0	0.040	0.030	2.0 to 3.0
	—	—	A 376	—	0.04 to 0.10	2.00	0.75	16.0 to 18.0	11.0 to 14.0	0.030	0.030	2.0 to 3.0
2 ¹ / ₄ Cr-1Mo	2 ¹ / ₄ Cr- 1 Mo	—	—	A 387 Grade D	0.15 max	0.40 to 0.65	0.50	2.00 to 2.50	—	0.035	0.035	0.90 to 1.10
	—	—	A 355 P22	—	0.15 max	0.30 to 0.60	0.50	1.90 to 2.60	—	0.030	0.030	0.87 to 1.13

Note: Cobalt not to exceed 0.1% for primary piping.

TABLE 2.2 — EFFECT OF GRAIN SIZE ON THE ROOM-
TEMPERATURE MECHANICAL PROPERTIES OF ANNEALED
STAINLESS STEEL BARS CONTAINING
0.07% C, 18.8% Cr, AND 8.6% Ni^{7,8}

<u>Mechanical Properties</u>	<u>ASTM Grain Size</u>	
	<u>Fine</u> <u>(5 to 7)</u>	<u>Coarse</u> <u>(0 to 1)</u>
Tensile strength, psi	99,000	91,000
Yield strength (0.5% extension), psi	39,000	32,000
Elongation (in 2 in.), %	72	76
Brinell hardness	166	156

TABLE 2.3 — HEAT TREATMENT SCHEDULES OF THE PRIMARY MATERIALS⁵

<u>Steel</u>	<u>Hot Bending Temperature</u>	<u>Forging Temperature</u>	<u>Recommended Post-Forming Heat Treatment</u>	<u>Other Heat Treatment</u>
304 SS	Slow heating to 1500 to 1600°F, hot forming at 2100 to 2300°F	See hot bending	Reheat to 1900 to 2000°F, fast cool (water quench if possible)	Solution anneal (see column on left)
316 SS	As above	As above	As above except max- imum temperature 2050°F	As above
2 ¹ / ₄ Cr-1Mo	1750 to 1850°F	2000 to 2200°F	<u>Stress relieve</u> at 1300 to 1400°F and air cool, or <u>Anneal</u> at 1600°F and slow cool	Normalize at 1650 to 1750°F and temper at 1200 to 1400°F

TABLE 2.4 — TRANSFORMATION
TEMPERATURES, $2\frac{1}{4}$ Cr-1 Mo⁵

<u>On Heating 250°F/hr</u>		<u>On Cooling 50°F/hr</u>	
<u>Ac₁</u>	<u>Ac₃</u>	<u>Ar₃</u>	<u>Ar₁</u>
1480	1600	1525	1325

TABLE 2.5 — MODULUS OF ELASTICITY VS TEMPERATURE

Temperature, °F	Modulus of Elasticity, psi x 10 ⁻⁶	
	304 SS and 316 SS ^{47,42}	2 ¹ / ₄ Cr-1 Mo ^{47,42}
70	28.3	29.9
100	28.1	29.8
200	27.7	29.5
300	27.1	29.0
400	26.6	28.6
500	26.1	28.0
600	25.4	27.4
650	25.1	27.1
700	24.8	26.6
750	24.4	26.3
800	24.1	26.0
850	23.7	25.6
900	23.3	25.3
950	22.9	24.9
1000	22.5	24.5
1050	22.1	24.1
1100	21.7	23.7
1150	21.3	
1200	20.9	
1250	20.5	
1300	20.1	
1350	19.7	
1400	19.2	

TABLE 2.6 — SHEAR MODULUS VS TEMPERATURE

Temperature, °F	Shear Modulus, psi x 10 ⁻⁶	
	304 SS and 316 SS ⁴⁷	2 $\frac{1}{4}$ Cr-1 Mo ⁴⁷
70	11.2	12.0
100	11.1	11.9
200	10.9	11.8
300	10.6	11.6
400	10.4	11.3
500	10.1	11.0
600	9.9	10.7
650	9.7	10.6
700	9.6	10.4
750	9.5	10.3
800	9.3	10.1
850	9.2	9.9
900	9.1	9.8
950	8.9	9.6
1000	8.8	9.4
1050	8.6	9.2
1100	8.5	9.0
1150	8.3	
1200	8.2	
1250	8.0	
1300	7.9	
1350	7.7	
1400	7.6	

TABLE 2.7 — POISSON'S RATIO VS TEMPERATURE

Temperature, °F	<u>Poisson's Ratio vs Temperature</u>	
	<u>304 SS and 316 SS⁴⁷</u>	<u>2$\frac{1}{4}$ Cr-1 Mo⁴⁷</u>
70	0.24	0.26
100	0.25	0.27
200	0.27	0.28
300	0.29	0.29
400	0.30	0.30
500	0.31	0.30
600	0.32	0.30
650	0.32	0.30
700	0.32	0.30
750	0.32	0.30
800	0.32	0.30
850	0.32	0.30
900	0.32	0.29
950	0.31	0.29
1000	0.31	0.29
1050	0.31	0.28
1100	0.30	0.28
1150	0.30	
1200	0.29	
1250	0.29	
1300	0.28	
1350	0.27	
1400	0.27	

TABLE 2.8 — DENSITY VS TEMPERATURE

Temperature, °F	Density, lb/in. ³		
	304 SS ^{152, 153}	316 SS ^{152, 153}	2 ¹ / ₄ Cr-1 Mo *
70	0.290	0.288	0.282
100	0.290	0.288	0.282
200	0.289	0.287	0.281
300	0.288	0.286	0.281
400	0.287	0.285	0.280
500	0.286	0.284	0.280
600	0.285	0.283	0.279
700	0.285	0.283	0.278
800	0.284	0.282	0.277
900	0.283	0.281	0.277
1000	0.282	0.280	0.276
1100	0.281	0.279	0.275
1200	0.280	0.278	
1300	0.279	0.277	
1400	0.278	0.276	

*Values interpolated from density data for similar alloys from Reference 49.

TABLE 2.9 — MEAN COEFFICIENT OF THERMAL EXPANSION*
VS TEMPERATURE

Temperature, °F	Mean Coefficient of Thermal Expansion, in./in.-°F × 10 ⁶	
	304 SS and 316 SS ⁴⁰	2 ¹ / ₄ Cr-1 Mo ⁴⁰
70	9.11	6.07
100	9.16	6.13
200	9.34	6.38
300	9.47	6.60
400	9.59	6.82
500	9.70	7.02
600	9.82	7.23
650	9.87	7.33
700	9.93	7.44
750	9.99	7.54
800	10.05	7.65
850	10.10	7.74
900	10.16	7.84
950	10.22	7.91
1000	10.29	7.97
1050	10.34	8.05
1100	10.39	8.12
1150	10.43	
1200	10.48	
1250	10.51	
1300	10.54	
1350	10.57	
1400	10.60	

*The mean coefficient of thermal expansion at temperature T,
is given by:

$$\frac{\text{Length at } T - \text{length at } 70^{\circ}\text{F}}{(\text{Length at } 70^{\circ}\text{F}) (T - 70^{\circ}\text{F})}$$

TABLE 2.10 — INSTANTANEOUS COEFFICIENT OF THERMAL EXPANSION

Temperature, °F	Instantaneous Coefficient of Thermal Expansion, in./in.-°F × 10 ⁶	
	304 SS and 316 SS ^{153,154,42}	2 ¹ / ₄ Cr-1 Mo ⁴²
70	9.11	6.07
100	9.21	6.20
200	9.50	6.67
300	9.73	7.10
400	9.96	7.54
500	10.20	7.96
600	10.43	8.35
650	10.54	8.54
700	10.66	8.76
750	10.81	8.94
800	10.90	9.16
850	11.00	9.34*
900	11.09	9.52*
950	11.18	9.70*
1000	11.28	9.88*
1050	11.38	10.05*
1100	11.47	10.20*
1150	11.56	
1200	11.65	
1250	11.75	
1300	11.84	
1350	11.93	
1400	12.01	

*Extrapolated from Reference 42.

TABLE 2.11 — THERMAL CONDUCTIVITY VS TEMPERATURE

Temperature, °F	Thermal Conductivity, Btu/hr-ft-°F		
	304 SS and 316 SS ⁴²	2 $\frac{1}{4}$ Cr-1 Mo*	2 $\frac{1}{4}$ Cr-1 Mo†
70	8.35	16.25	36.00
100	8.40	16.25	36.60
200	8.90	16.25	34.85
300	9.35	16.25	33.20
400	9.80	16.25	31.60
500	10.23	16.25	29.80
600	10.27	16.25	28.30
650	10.29	16.25	27.50
700	11.10	16.25	26.80
750	11.35	16.25	26.00
800	11.55	16.25	25.40
850	11.80	16.25	24.60
900	11.90	16.25	23.80
950	12.20	16.25	23.20
1000	12.42	16.25	22.60
1050	12.70	16.25	22.00
1100	12.90	16.25	21.40
1150	13.10		
1200	13.30		
1250	13.55		
1300	13.80		
1350	14.00		
1400	14.20		

*Values taken from References 37, 161, and 5 specifically for 2 $\frac{1}{4}$ Cr-1 Mo.

†Values taken from Reference 42 which lists a single set of values for
“carbon steel, carbon-moly steel” and “low-chrome steels (through 3% Cr).”

TABLE 2.12 — SPECIFIC HEAT VS TEMPERATURE

Temperature, °F	Specific Heat, Btu/lb-°F	
	304 SS and 316 SS ^{152, 153}	2 ¹ / ₄ Cr-1 Mo*
70	0.108	0.107
100	0.110	0.108
200	0.116	0.111
300	0.120	0.115
400	0.124	0.119
500	0.126	0.124
600	0.128	0.131
650	0.129	0.134
700	0.130	0.138
750	0.131	0.141
800	0.132	0.145
850	0.133	0.150
900	0.134	0.154
950	0.135	0.158
1000	0.137	0.163
1050	0.138	0.168
1100	0.139	0.172
1150	0.140	
1200	0.141	
1250	0.143	
1300	0.144	
1350	0.145	
1400	0.147	

*Calculated by averaging values of specific heat for the elements present in nominal composition 2¹/₄ Cr-1 Mo steel using specific heat data of Reference 155.

TABLE 2.13 — THERMAL DIFFUSIVITY VS TEMPERATURE

Temperature, °F	Thermal Diffusivity, ft ² /hr	
	304 SS and 316 SS ⁴²	2 $\frac{1}{4}$ Cr-1 Mo*
70	0.150	0.421
100	0.150	0.417
200	0.155	0.408
300	0.159	0.393
400	0.163	0.382
500	0.166	0.366
600	0.171	0.348
650	0.172	0.341
700	0.174	0.331
750	0.176	0.325
800	0.178	0.317
850	0.180	0.306
900	0.181	0.298
950	0.183	0.291
1000	0.185	0.283
1050	0.188	0.275
1100	0.190	0.269
1150	0.192	
1200	0.193	
1250	0.196	
1300	0.198	
1350	0.200	
1400	0.202	

*Calculated from data of Table 2.8, Column 3; Table 2.11, Column 2; and Table 2.12, Column 2.

TABLE 2.14 — AVERAGE ELONGATION OF
PRIMARY MATERIALS VS TEMPERATURE⁵

Tempera- ture, °F	Elongation, % in 2 in.		
	<u>304 SS</u>	<u>316 SS</u>	<u>2¹/₄ Cr-1 Mo</u>
70	63	60	43
300	60	53	35
500	55	49	27
700	56	47	24
900	45	45	30
1100	40	43	44
1300	37	42	64

TABLE 2.15 — EFFECTS OF AGING 304 SS AT ELEVATED TEMPERATURES ON IMPACT STRENGTH, HARDNESS, TENSILE PROPERTIES, AND RUPTURE PROPERTIES⁵

Effect of prolonged exposure at elevated temperatures⁽¹⁾

Impact strength and hardness

	Un- (3) exposed	Exposed 1000 hrs at			Exposed 10,000 hrs at		
		900F	1050F	1200F	900F	1050F	1200F
Room Temperature Charpy Keyhole Notch Impact Values (Ft-Lbs)	91	87	75	60	79	62	47
Brinell Hardness	141	145	142	143	143	152	143
Transition Temp F (2)	<-420	..	..	..	<-300	<-300	<-300

Room temperature tensile properties

10,000 Hour Exposure Temperature F	0.2% Offset Yield Strength 1000 psi	Tensile Strength 1000 psi	Elongation % in 1 in	Reduction of Area %
Unexposed	32.2	81.5	72	80
900	31.6	82.8	68	80
1050	35.9	86.8	63	76
1200	29.4	80.8	63	69

Rupture properties

Exposure and Testing Temperature F	Stress (1000 psi) for Rupture in					
	1000 hrs	10,000* hrs	100,000* hrs	1000 hrs	10,000* hrs	100,000* hrs
	Before Exposure (3)			After Exposure Without Stress for 10,000 hrs.		
900	49.0	38.0	29.7	55.0		
1050	27.0	21.5	16.3	29.5	22.0	16.5
1200	14.0	9.9	6.9	17.2	12.3	9.1

Notes: (1) These data represent tests conducted on one heat of steel.

(2) Selected at the 15 ft-lb level.

(3) Prior heat treatment: 1900°F, 1/2 hr, WQ.

*Values obtained by extrapolation.

TABLE 2.16 — EFFECTS OF AGING 316 SS AT ELEVATED TEMPERATURES ON IMPACT STRENGTH, HARDNESS, TENSILE PROPERTIES, AND RUPTURE PROPERTIES⁵

Effect of prolonged exposure at elevated temperatures (1)

Impact strength and hardness

	Un- exposed	(3) Exposed 1000 hrs at			Exposed 10,000 hrs. at		
		900F	1050F	1200F	900F	1050F	1200F
Room Temperature							
Charpy Keyhole Notch							
Impact Values (Ft-Lbs)	80	86	72	44	87	43	21
Brinell Hardness	143	151	148	170	145	163	177
Transition Temp F (2)	< -320				< -300	-230	...

Room temperature tensile properties

10,000 Hour Exposure Temperature F	0.2% Offset Yield Strength 1000 psi	Tensile Strength 1000 psi	Elongation % in 1 in	Reduction of Area %
Unexposed	37.9	88.4	67	78
900	43.0	88.7	62	78
1050	38.2	92.2	57	69
1200	42.5	100.1	42	58

Rupture properties

Exposure and Testing Temperature F	Stress (1000 psi) for Rupture in					
	1000 hrs	10,000* hrs.	100,000* hrs	1000 hrs	10,000* hrs	100,000* hrs
	Before Exposure (3)			After Exposure Without Stress for 10,000 hrs		
900	67.0	...	...	71.0	...	...
1050	40.2	32.0	25.5	44.7	37.0	30.5
1200	23.0	16.3	11.7	22.3	16.7	12.3

Notes: (1) These data represent tests conducted on one heat of steel.

(2) Selected at the 15 ft-lb level

(3) Prior Heat treatment: 1950°F, 1/2 hr, WQ.

*Values obtained by extrapolation.

TABLE 2.17 — EFFECTS OF AGING $2\frac{1}{4}$ Cr-1 Mo AT ELEVATED TEMPERATURES ON IMPACT STRENGTH, HARDNESS, TENSILE PROPERTIES, AND RUPTURE PROPERTIES⁵

Effect of prolonged exposure at elevated temperatures⁽¹⁾

Impact strength and hardness

	Un- ⁽²⁾ exposed	Exposed 1000 hrs. at			Exposed 10,000 hrs. at		
		900F	1050F	1200F	900F	1050F	1200F
Room Temperature							
Charpy Keyhole Notch							
Impact Values (Ft-Lbs)	73	63	74	71	81	69	77
Brinell Hardness	130	163	119	109	149	143	137

Room temperature tensile properties

10,000 Hour Exposure Temperature F	0.2% Offset Yield Strength 1000 psi	Tensile Strength 1000 psi	Elongation % in 1 in.	Reduction of Area %
Unexposed	57.1	80.4	32	80
900	40.5	72.3	34	79
1050	41.3	70.6	36	75
1200	41.3	64.9	37	77

Rupture properties

Exposure and Testing Temperature F	Stress (1000 psi) for Rupture in					
	1000 hrs.	10,000* hrs.	100,000* hrs.	1000 hrs.	10,000* hrs.	100,000* hrs.
	Before Exposure ⁽²⁾			After Exposure Without Stress for 10,000 hrs		
900	45.5	36.4	29.3	33.5	29.0	24.8
1050	24.5	17.5	12.4	16.0	12.0	9.9
1200	7.5	4.2	2.4	6.1	4.3	2.9

Notes: (1) These data represent tests conducted on one heat of steel.

(2) Prior heat treatment: normalized at 1650°F, tempered at 1300°F.

*Values obtained by extrapolation.

TABLE 2.18 — TENSILE PROPERTIES OF ANNEALED TYPES 304, 316, AND 348 STAINLESS STEEL SPECIMENS TESTED IN AIR¹¹⁵

Material	Testing Temperature		Strain Rate, sec ⁻¹	Tensile Modulus, 10 ⁶ psi	Poisson's Ratio	Tensile Strength, psi	Reduction in Area, percent	Tensile Ductility, cf, percent
	C	F						
304	21	70	4 x 10 ⁻³	28.7	0.264	89,000	80.6	164.0
304	21	70	4 x 10 ⁻⁴	28.7	0.264	--	--	--
304	21	70	4 x 10 ⁻⁵	28.7	0.264	--	--	--
304	430	806	4 x 10 ⁻³	23.4	0.282	62,400	64.0	102.2
304	430	806	4 x 10 ⁻⁴	23.4	0.282	--	--	--
304	430	806	4 x 10 ⁻⁵	23.4	0.282	64,600	64.5	103.6
304	650	1202	4 x 10 ⁻³	21.6	0.315	45,600	42.2	54.8
304	650	1202	4 x 10 ⁻⁴	21.6	0.315	--	--	--
304	650	1202	4 x 10 ⁻⁵	21.6	0.315	34,500	33.4	40.6
304	816	1500	4 x 10 ⁻³	18.8	0.323	25,000	51.0	71.3
304	816	1500	4 x 10 ⁻⁴	18.8	0.323	--	--	--
304	816	1500	4 x 10 ⁻⁵	18.8	0.323	13,400	32.1	38.7
348	21	70	4 x 10 ⁻³	28.2	0.250	--	--	--
348	21	70	4 x 10 ⁻⁴	28.2	0.250	--	--	--
348	21	70	4 x 10 ⁻⁵	28.2	0.250	--	--	--
348	430	806	4 x 10 ⁻³	23.8	0.275	59,600	66.4	109.1
348	430	806	4 x 10 ⁻⁴	23.8	0.275	--	--	--
348	430	806	4 x 10 ⁻⁵	23.8	0.275	62,600	57.5	85.6
348	650	1202	4 x 10 ⁻³	21.8	0.295	47,700	68.6	115.8
348	650	1202	4 x 10 ⁻⁴	21.8	0.295	--	--	--
348	650	1202	4 x 10 ⁻⁵	21.8	0.295	40,800	40.4	51.8
348	816	1500	4 x 10 ⁻³	19.05	0.340	25,900	87.0	204.0
348	816	1500	4 x 10 ⁻⁴	19.05	0.340	--	--	--
348	816	1500	4 x 10 ⁻⁵	19.05	0.340	17,000	84.0	183.3
316	21	70	4 x 10 ⁻³	30.1	0.295	90,000	74.6	137.0
316	21	70	4 x 10 ⁻⁴	30.1	0.295	--	--	--
316	21	70	4 x 10 ⁻⁵	30.1	0.295	--	--	--
316	430	806	4 x 10 ⁻³	24.0	0.315	67,700	62.1	97.0
316	430	806	4 x 10 ⁻⁴	24.0	0.315	--	--	--
316	430	806	4 x 10 ⁻⁵	24.0	0.315	73,900	60.6	93.1
316	650	1202	4 x 10 ⁻³	21.95	0.326	54,700	61.1	94.0
316	650	1202	4 x 10 ⁻⁴	21.95	0.326	--	0-	--
316	650	1202	4 x 10 ⁻⁵	21.95	0.326	42,800	32.1	38.7
316	816	1500	4 x 10 ⁻³	18.4	0.321	29,700	61.6	95.7
316	816	1500	4 x 10 ⁻⁴	18.4	0.321	--	--	--
316	816	1500	4 x 10 ⁻⁵	18.4	0.321	21,300	49.9	69.1

TABLE 2.19 — CHARPY IMPACT PROPERTIES
OF $2\frac{1}{4}$ Cr-1 Mo ALLOY STEEL AT ELEVATED
TEMPERATURES³⁷ IN FT-LB*

Temperature, °F	$2\frac{1}{4}$ Cr-1 Mo
RT	59
200	64
300	64
400	53
500	54
600	53
700	40
800	35
900	33
1000	30
1100	30
1200	34
1300	41

*All values are the average of duplicate tests on
Standard 0.394-in. square, keyhole notch Charpy
Specimens taken from tubes or bars with the notch
normal to fiber direction.

TABLE 2.20 — EFFECT OF ELEVATED TEMPERATURE EXPOSURE ON IMPACT STRENGTH²²

			Room-Temperature Charpy Keyhole-Notch Impact, ft-lb							
			after Exposure at							
Type	Chromium Equivalent	Original	1150°F (620°C)		1250°F (675°C)		1350°F (730°C)		1450°F (790°C)	
			10,000 hr	100,000 hr	10,000 hr	100,000 hr	10,000 hr	100,000 hr	10,000 hr	100,000 hr
316	19.7	83.5	32	18	—	—	—	—	—	—
316	20.1	52	13	6.5	—	—	—	—	10	5
316	20.2	97.5	36	24	14.5	4	—	—	—	—
316	20.5	105	19	4	—	—	10	4	—	—
316	20.6	99.5	—	—	11	3	9	4	—	—
316	20.7	72	—	—	—	—	6	2	—	—
316	21.9	81.5	26	8	25	9	6.5	2	—	—
316 L	19.4	78.5	40	33	35	11	31	10	19	6

TABLE 2.20 — EFFECT OF ELEVATED TEMPERATURE EXPOSURE ON IMPACT STRENGTH²²

			Room-Temperature Charpy Keyhole-Notch Impact, ft-lb							
			after Exposure at							
Type	Chromium Equivalent	Original	1150°F (620°C)		1250°F (675°C)		1350°F (730°C)		1450°F (790°C)	
			10,000 hr	100,000 hr	10,000 hr	100,000 hr	10,000 hr	100,000 hr	10,000 hr	100,000 hr
316	19.7	83.5	32	18	—	—	—	—	—	—
316	20.1	52	13	6.5	—	—	—	—	10	5
316	20.2	97.5	36	24	14.5	4	—	—	—	—
316	20.5	105	19	4	—	—	10	4	—	—
316	20.6	99.5	—	—	11	3	9	4	—	—
316	20.7	72	—	—	—	—	6	2	—	—
316	21.9	81.5	26	8	25	9	6.5	2	—	—
316 L	19.4	78.5	40	33	35	11	31	10	19	6

TABLE 2.21 — TYPICAL ENDURANCE LIMITS
FOR ANNEALED SPECIMENS OF 304 SS, 316 SS,
AND $2\frac{1}{4}$ Cr-1 Mo AT 80°F

<u>Steel</u>	<u>Endurance Limit, psi</u>
304 SS	35,000 ⁽⁵⁾
316 SS	39,000 ⁽⁵⁾
$2\frac{1}{4}$ Cr-1 Mo	32,000 (estimated from Fig. 2.14)

TABLE 2.22 — AVERAGE STRESS FOR RUPTURE IN 100,000 HR-1000 PSI
(BY LINEAR REGRESSION OF LOG STRENGTH VS TEMPERATURE)

Temp, °F	<u>304L</u>	<u>304- 304H</u>	<u>304-304H*</u> <u>Adjust.</u>	<u>316L</u>	<u>316- 316L</u>	<u>316-316H*</u> <u>Adjust.</u>	<u>321</u>	<u>321*</u> <u>Adjust.</u>	<u>321H</u>	<u>347- 347H</u>	<u>347-347H*</u> <u>Adjust.</u>
950											
1000	19.5†	25.8	29.0	34.5†	37.0†	—	—	—	29.0†	37.5	40.0
1050	15.0	20.0	22.6	25.0	28.0	28.5	23.0†	22.7	22.5	28.0	29.8
1100	11.6	15.8	17.8	18.5	20.8	21.2	16.5	16.3	17.4	20.8	22.2
1150	8.9	12.5	14.1	13.7	15.1	15.35	12.0	11.9	13.3	15.3	16.3
1200	6.9	9.8	11.04	10.1	11.4	11.6	8.7	8.6	10.3	11.5	12.25
1250	5.3	7.6	8.6	7.4	8.4	8.55	6.3	6.23	7.9	8.6	9.15
1300	4.1	6.0	6.8	5.5	6.3	6.4	4.6	4.55	6.1	6.4	6.8
1350	3.15	4.7	5.3	4.0	4.7	4.8	3.3	3.26	4.7	4.7	5.0
1400	2.4	3.7	4.17	3.0	3.45	3.51	2.45	2.42	3.6†	3.55†	3.78
1450	1.87	2.9	3.27	2.2	2.6	2.64	1.75	1.73	2.8†	2.64†	2.81
1500	1.45	2.3	2.59	1.6	1.95	1.98	1.27	1.26	2.12†	1.95†	2.08

*By the percentage required to adjust the value at 1200°F, from regression analysis, to equal the average of the values at 1200°F developed by visual extrapolation of individual lots. In view of the small number of data, no adjustment was attempted for 304L and 316L.

†By extrapolation to temperatures outside of test range.

TABLE 2.23 — MINIMUM STRESS FOR RUPTURE IN 100,000 HR — 1000 PSI

Temp °F	304-304H		316-316H		321		321H		347-347H	
	1	2	1	2	1	2	1	2	1	2
950										
1000	18.1	19.9					21.5	18.8	32.7	32.2
1050	14.1	15.5	23.1	22.8	14.52	12.0	16.7	14.6	24.3	23.9
1100	11.1	12.2	17.15	16.9	10.43	8.6	12.9	11.3	18.1	17.8
1150	8.8	9.65	12.4	12.25	7.62	6.28	9.85	8.62	13.3	13.1
1200	6.9	7.55	9.4	9.28	5.5	4.53	7.63	6.68	10.0	9.85
1250	5.37	5.9	6.91	6.82	3.99	3.29	5.85	5.12	7.47	7.35
1300	4.25	4.65	5.18	5.11	2.91	2.40	4.52	3.96	5.55	5.45
1350	3.31	3.63	3.88	3.84	2.09	1.72	3.48	3.05	4.08	4.02
1400	2.61	2.86	2.84	2.80	1.55	1.28	2.67	2.33	3.09	3.04
1450	2.04	2.24	2.14	2.11	1.11	.91	2.03	1.82	2.29	2.26
1500	1.62	1.77	1.60	1.58	0.81	.66	1.57	1.37	1.70	1.69

1. Computed by multiplying average trend curve (adjusted when necessary) by ratio of observed individual minimum value at 1200°F to average at 1200°F.
2. Computed by multiplying average trend curve (adjusted when necessary) by ratio of statistically defined minimum (average less 1.65 times the standard deviation) at 1200°F to the average at 1200°F.

TABLE 2.24 — STRESS TO PRODUCE RUPTURE IN 200,000 HR, BASED
ON AVERAGE DATA VS TEMPERATURE

Temperature, °F	Rupture Strength, psi ¹⁴⁵				
	304H	304L	316H	316L	2½Cr-1 Mo
800	--	--	--	--	--
850	--	--	--	--	--
900	--	--	--	--	--
950	--	--	--	--	14,250
1000	23,000	18,000	--	33,600	10,900
				min 27.0	
1050	18,000	13,900	27,000	23,500	7,500
			min 19.8	min 21.0	
1100	14,000	10,700	20,000	17,000	6,000
			min 14.8	min 15.7	
1150	11,200	8,000	14,000	12,300	3,950
			min 10.9	min 11.6	
1200	8,900	6,250	10,300	9,100	--
			min 8.0	min 8.5	

TABLE 2.25 — STRESS TO RUPTURE IN 20,000 HR, BASED
ON MINIMUM DATA, VS TEMPERATURE

Temperature, °F	Rupture Strength, psi ¹⁴⁵			
	304H	304L	316H	316L
800	--	--	--	--
850	--	--	--	--
900	--	--	--	--
950	--	--	--	--
1000	17,250	14,200	--	27,000
1050	13,500	11,300	19,800	21,000
1100	10,500	8,900	14,800	15,700
1150	8,400	7,000	10,900	11,600
1200	6,600	5,500	8,000	8,500

Note: 200,000-hr data extrapolated from 10,000-hr and
100,000-hr data.

TABLE 2.26 — CREEP STRESS TO PRODUCE 0.005% STRAIN IN 1000 HR
VS TEMPERATURE

Temperature, °F	Creep Stress, psi ¹⁴⁵				
	<u>304H</u>	<u>304L</u>	<u>316H</u>	<u>316L</u>	<u>2¼ Cr-1 Mo</u>
800	--	--	--	--	--
850	--	12,800	--	--	--
900	--	10,500	--	--	22,000
950	--	8,500	--	--	--
1000	16,000	7,000	17,000	22,000	7,500
1050	--	5,600	13,000	16,000	--
1100	10,000	4,500	10,400	11,500	4,000
1150	--	3,500	8,250	8,400	--
1200	6,300	2,800	6,600	5,900	2,500

TABLE 2.27 — THERMAL EXPANSION COEFFICIENT
(in./in./°F $\times 10^{-6}$)

<u>Alloy</u>	<u>70°F/212°F</u>	<u>70°F/600°F</u>
304 SS	9.6	9.9
316 SS	8.9	9.8
309 SS	8.3	9.3
2 $\frac{1}{4}$ Cr-1 Mo	6.7	7.2
Inconel 82/182	7.6	8.3

TABLE 2.28 — MECHANICAL TEST RESULTS AND ALL-WELD-METAL
CHEMICAL ANALYSIS — FINAL 4½-IN. THICK TRANSITION-WELD TEST PLATE
(2¼ Cr-1 Mo to 316 SS) ⁴⁶

Side Bend Tests

Six guided side-bend specimens bent 180° without defects.

Reduced Section Tensile Tests

<u>Specimen</u>	<u>Yield, psi</u>	<u>Tensile, psi</u>	<u>% Elongation in 2 in.</u>	<u>Fracture</u>
Top	49,840	78,830	54.5	Stainless base metal
Center	49,190	76,620	60.0	Stainless base metal
Bottom	47,880	77,610	48.5	Stainless base metal
Average	48,970	77,680	54.5	

ASTM Minimum Requirements

A182-58T-F316	30,000	75,000	40
A182-58T-F22	40,000	70,000	20
A371-53T-309	—	80,000	35

Chemical Analysis of All-Weld Metal (309 SS)

<u>C, %</u>	<u>Mn, %</u>	<u>Si, %</u>	<u>P, %</u>	<u>S, %</u>	<u>Cr, %</u>	<u>Ni, %</u>	<u>Delta Ferrite</u>
0.05	1.50	1.02	0.031	0.023	22.37	13.90	6.5%

Fig. 2.1 — Effect of Time and Temperature on the Maximum Sensitivity to Intergranular Attack of Low-Carbon Steel Containing 18% Cr, 8% Ni (Reference 190)

Fig. 2.2 — The Influence of Carbon Content and Temperature of Heat-Treatment on the 1000-hr Rupture Strength at 1200°F of Laboratory Heats of Type 304 Steel Containing very Low Nitrogen ($<0.01\%$)¹¹⁷

Fig. 2.3 — Diffusion Coefficient of Carbon in Austenite³³

Fig. 2.4 — The Influence of Nitrogen at Various Carbon Levels on the 1000-hr Rupture Strength at 1200°F of wholly Austenitic Type 304 Steel. The Values Near the Symbols Are the Estimated Percent Elongation on Rupture at 1000 hr.¹¹⁷

Extrapolated 100,000-Hour Rupture Strength
at 1200°F, ksi

Data Point	Composition				Heat Treatment				Heat Treat Temp °F	Extrapolated 100,000-Hr Rupture Strength ksi	Notes
	C	N	Mn	P	C	N	Mn	P			
1	0.05	0.053	0.28	0.04	10.12	1.35	0.121	H V	170	((6.3)) ^a	-
2	0.03	0.031	0.5	0.04	10.22	1.55	0.03	H W	1700	((8.3))	8
3	0.05	0.011	0.5	0.1	10.12	1.55	0.1	H	1700	1	5
4	0.05	0.034	<0.1	0.1	10.70	1.85	0.05	H V	2050	8.0	9
5	0.050	0.10	1.4	0.14	1.2	1.13	0.155	H W	17.5	((8.3))	12
6	0.025	0.13	1.3	0.55	8	18.87	0.15	H W	17.5	((7.3))	1
7	0.0	0.04	1.2	0.4	1.51	18.6	0.170	H W	17.5	((7.3))	1
8	0.0	0.026	0.53	0.01	10.1	18.53	0.051	H W	196	2	14
9	0.018	0.14	-	-	11.3	13.2	0.143	-	-	((10.5))	1
10	0.018	0.032	-	-	11.3	18.1	0.08	-	-	((8.5))	1
11	0.052	0.10	1.6	0.55	10.40	18.1	0.1	H W	17.5	((10.8))	1
12	0.052	0.1	1.7	0.4	10.33	1.67	0.181	C W	175	11.3	1
13	0.050	0.10	1.4	0.3	10.45	18.80	0.1	C W	17	((11.0))	1
14	0.05	0.09	1.63	0.48	10.61	19.15	0.10	H V	1750	((11.0))	1
15	0.054	0.087	1.79	0.48	10.11	17.09	0.10	C W	2050	11.5	3
16	0.0	0.078	1.35	0.12	1.0	18.50	0.075	C W	2050	8	5
17	0.04	0.031	1.47	0.50	10.00	17.4	0.0	C V	1850	5	3
18	0.0	0.0	0.61	0.50	1	1	0.1	C V	2050	9.7	3
19	0.00	0.04	2.0	0.4	1.3	18.4	0.1	C W	20	9.4	3
20	0.02	0.045	1.10	0.32	10.50	18.55	0.0	H V	195	8	2
21	0.020	0.13	1.12	0.31	9.48	18.17	0.185	H W	-	10.0	2
22	0.033	0.030	<0.1	<0.1	10.42	18.8	0.073	H W	2000	((6.2)) ^b	10
23	0.03	0.012	<0.1	<0.1	11.18	1.2	0.10	H	200	((7.8)) ^b	1
24	0.101	0.032	<0.1	<0.1	10.7	18.26	0.153	H W	2000	((6.4)) ^b	10

Notes: 1. Values are for the present test.

Point No.	Heat	Point No.	Heat	Point No.	Heat	Point No.	Heat
25	15.5	30	17	3	14105	40	13.5
26	1711	31	131	36	13.3	41	13111
27	1557	32	147 A	37	13331	42	13477
28	1781	33	1.13	38	14173	43	13053
29	1262	3	13.6	40	13726		

Notes:

a - Type of heat treatment
H V - Heat V
C W - Cold Water

b - Values are for the present test. Values in parentheses are for data
c - Values are for the present test. Values in parentheses are for data

Fig. 2.5 — Influence of Carbon and Nitrogen on 100,000-hr Rupture Strength at 1200°F of wholly Austenitic Type 304 Steel; Composition and History of Materials Given in Table. (Most of data in lower shaded region are interpolated or are from extra low manganese experimental material.)¹¹⁷

Fig. 2.6 — Diffusion Coefficient of Carbon in Alpha Iron¹¹²

Fig. 2.7 — Effect of Chromium Content on the Activity of Carbon in the Iron-Chromium-Carbon System at 1000°C¹⁴⁴

Fig. 2.8 — Average Instantaneous Coefficients of Expansion of Chromium-Nickel Stainless Steel and 2 $\frac{1}{4}$ Cr-1 Mo^{8,37}

Fig. 2.9 — Mean Coefficients of Thermal Expansion of 304SS, 316 SS, and 2 $\frac{1}{4}$ Cr-1 Mo between 70 $^{\circ}\text{F}$ and the Indicated Temperatures^{8,37}

Fig. 2.10 — Effect of Temperature on the Thermal Conductivity of Chromium-Nickel Stainless and 2 1/4 Cr-1 Mo Steels^{8,37}

Annealed Bar, Plate, Heavy - Wall Pipe

Fig. 2.11 — Range of Short-Time Tensile Properties of Annealed Type 304 Stainless Steel⁵⁰⁻⁵⁵

a. Bar and Heavy - Wall Pipe

b. Bar, Plate, and Heavy - Wall Pipe

Fig. 2.12 — Short-Time Tensile Properties of Annealed Type 316 Stainless Steel^{5, 50, 52, 54, 55, 62}

(Tensile Strength - 2.25 Cr-1 Mo Steel)

(0.2% Offset Yield Strength - 2.25 Cr-1 Mo Steel)

(% Elongation - 2.25 Cr-1 Mo Steel)

Fig. 2.13 — Range of Tensile Properties vs Temperature for Annealed $2\frac{1}{4}$ Cr-1 Mo Steel⁵⁷

a. Type 304-304H Bar

b. Type 304-304H Pipe-Tube

Fig. 2.14 — Variation of Yield Strength Ratio with Temperature¹⁴⁵

c. Type 304-304H Plate

d. Type 304L

Fig. 2.14 — Variation of Yield Strength Ratio with Temperature (Continued),¹⁴⁵

e. Type 316

f. Type 316L

Fig. 2.14 — Variation of Yield Strength Ratio with Temperature (Continued)¹⁴⁵

a. Type 304-304H Bar

b. Type 304-304H Pipe Tube

Fig. 2.15 — Variation of Tensile Strength Ratio with Temperature¹⁴⁵

c. Type 304-304H Plate

d. Type 304L

Fig. 2.15 — Variation of Tensile Strength Ratio with Temperature
(Continued)¹⁴⁵

e. Type 316

f. Type 316L

Fig. 2.15 — Variation of Tensile Strength Ratio with Temperature
(Continued)¹⁴⁵

Fig. 2.16 — True Stress-True Strain Curves for Tensile Tests in Air for Annealed Type 304 Stainless Steel at Constant Axial Strain Rates of 4×10^{-3} and $4 \times 10^{-5}/\text{sec}^{171}$

Bar Stock at Elevated Temperature

Fig. 2.17 — Charpy Keyhole Impact Strength of Type 304 Stainless Steel Bar at Elevated Temperatures^{50, 52, 53}

Fig. 2.18 — Charpy Keyhole Impact Strength of Type 316 Stainless Steel Bar at Elevated Temperatures^{50, 37}

Fig. 2.19 — Effect of Temperature and Time of Exposure on the Room-Temperature Impact Properties of Type 304⁸

Fig. 2.20 — Effect of Temperature and Time of Exposure on the Room-Temperature Impact Properties of Type 316⁸

Fig. 2.21 — Strain-Controlled Fatigue Results for 18-8 Stainless Steels at Various Temperatures⁵⁸

Fig. 2.23 — Fatigue Strength of 18-8 Stainless Steels at Various Elevated Temperatures⁵⁸

Fig. 2.24 — Fatigue Strength of 2 1/4 Cr-1 Mo Steel at Various Temperatures⁵⁸

Fig. 2.25 — Strain-Controlled Fatigue Results for 18-8 Stainless Steels (304, 348, 316 SS) at 800 F⁵⁸

Axial strain rate:

- (a) $\dot{\epsilon}_t = 4 \times 10^{-3} \text{ sec}^{-1}$
 (b) $\dot{\epsilon}_t = 4 \times 10^{-4} \text{ sec}^{-1}$
 (c) $\dot{\epsilon}_t = 4 \times 10^{-5} \text{ sec}^{-1}$

Temperature:

- 430°C (800°F)
 ● 650°C (1200°F)
 △ 816°C (1500°F)

Fatigue Life (N_5), Cycles

Fatigue Life (N_5), Cycles

Fig. 2.26 — Plastic Strain Range vs Fatigue Life for AISI 304 Stainless Steel at Various Temperatures

Axial strain rate:

- (a) $\dot{\epsilon}_t = 4 \times 10^{-3} \text{ sec}^{-1}$
 (b) $\dot{\epsilon}_t = 4 \times 10^{-4} \text{ sec}^{-1}$
 (c) $\dot{\epsilon}_t = 4 \times 10^{-5} \text{ sec}^{-1}$

Temperature:

- 430°C (800°F)
 ● 650°C (1200°F)
 △ 816°C (1500°F)

Fatigue Life (N_5), Cycles

Fatigue Life (N_5), Cycles

Fig. 2.27 — Plastic Strain Range vs Fatigue Life for AISI 316 Stainless Steel at Various Temperatures

Fig. 2.28 — Stress-Cycle Fatigue Results for Type 304 Stainless Steel Bar Tested at Various Temperatures (Rotating Beam: $R = -1.10$; $K_t = 1.0$)^{50, 59}

Fig. 2.29 — Predicted Fatigue Behavior for Annealed Type 304 Stainless Steel for Tests at 1200°F and a Strain Rate of $4 \times 10^{-3}/\text{sec}$ ¹⁷¹

a. Type 304-304H

b. Type 304 L

Fig. 2.30 — Variation of Rupture Strength (100,000 hr) and Creep Strength (0.01% per 1000 hr) with Temperature¹⁴⁵

c. Type 316-316H

d. Type 316L

Fig. 2.30 — Variation of Rupture Strength (100,000 hr) and Creep Strength (0.01% per 1000 hr) with Temperature (Continued)¹⁴⁵

Fig. 2.31a — Stress-Rupture Curve for 2 1/4 Cr-1 Mo (ASTM Avg Data)^{8,57}

Fig. 2.31b — Creep Rate Curve for 2 1/4 Cr-1 Mo (ASTM Avg Data)^{8, 57}

Fig. 2.32 — Stress vs Rupture-Time and Creep-Rate Curves for 2.25 Cr-1 Mo Steel Based on Average Data for Annealed Stock^{8, 57}

Fig. 2.33 — Plot for Determining Rupture Ductility - 304-304H SS Data¹²²

Fig. 2.34 — Isochronous Stress-Strain Curves Annealed
304-304H SS at 1100°F¹²¹

Fig. 2.35 — Time-Temperature Sensitization Relationship for 18-8 Weld Metal²⁵

Fig. 2.36 — Schaeffler's Diagram for the Microstructure of Stainless Steel Welds⁶¹

3. EFFECTS OF ENVIRONMENTAL CONDITIONS ON PIPING MATERIALS

3.1 INTRODUCTION

The composition and properties of the selected pipe materials, described in Section 2, may be significantly affected by the environmental conditions to which the LMFBR piping will be exposed. Of primary concern are the effects of sodium and its anticipated contaminants on corrosion, mass transfer, and mechanical properties. Also to be considered in other tasks are the effects of cover or purge gases, cleaning fluids, moisture, halides, fluxing agents external to the piping, and surrounding air or oxygen-depleted gases (nitrogen). Radiation effects are mentioned briefly.

The state-of-the-art of sodium effects may be characterized as being based on empirical information. Very little work has been done to explore systematically and/or fundamentally the effects of one environmental variable at a time. Sodium purity control and purity characterization often have been deficient. As a result, the information in this section presents a compilation of available data which in many instances may not be correlated.

Many of the data in sodium environments were obtained with thin-walled (typically 0.010 to 0.060 in.) test specimens. The use of thin specimens has the advantage of ascertaining in a relatively short test time whether the environment has any effect at all and what it might be. Both surface effects and effects due to slow diffusion of an element, e.g., carbon within the steel, are accentuated by the use of thin test specimens. However, tests with thin specimens do not provide the required design data for heavier section ($>1/2$ -in. thick) pipe, which will be used in LMFBR systems, therefore additional R&D is required to generate such data.

In spite of these shortcomings, certain major conclusions can be reached as follows:

1. The anticipated environmental effects are not so detrimental as to negate the use of the selected materials in LMFBR piping.
2. There are environmental effects which do cause corrosion and/or deterioration of mechanical properties to such an extent that they must be considered in selecting material composition, thickness, and allowable stresses for pipe design.

3. Material compositions, purities, and metallurgical factors must be specified and stringently controlled during all phases of fabrication, shipment, assembly, operation, maintenance, and repair to assure satisfactory performance.

The state-of-the-art of environmental effects, as presented in this Report 228, will be utilized in subsequent tasks to define materials design properties, corrosion allowances, and R&D requirements. It is expected that areas of major uncertainties will relate to carbon transfer and its long-term (>200,000 hr) effects on mechanical properties, surface removal (corrosion) rates in pipes with very large diameters (up to 48 in.), and possibly nitrogen depletion effects in pool-type systems. In addition, creep and fatigue properties are known to be sensitive to environmental effects. Where the available data are insufficient, judgments will have to be applied.

Erosion is not considered to be a significant factor in LMFBR piping because of the relatively low sodium velocities and the general freedom from particulate matter in the sodium. To ensure that it does not become a problem, it is important that conditions conducive to cavitation or vortex formation be avoided. Except for cavitation-induced erosion, severe sodium erosion has only been reported in two instances, both resulting from strong vortexing of the sodium in pump discharge headers.^{125,160} In one of these cases,¹²⁵ particulate matter suspended in the vortexing sodium was identified as causing the abrasion of the metal wall.

3.2 SODIUM CORROSION AND MASS TRANSFER

3.2.1 Mechanisms and Effects

Two basic mechanisms – solution attack and impurity (chemical) reactions – lead to corrosion of LMFBR piping surfaces by sodium. These mechanisms, individually or in combination, can cause mass transfer, which is the transport of material by the sodium from one point in the system to another. The resultant effects and engineering consequences may be depicted on the following page.⁷⁴

A. Solution Attack

Steel surfaces exposed to clean sodium will undergo solution attack until the sodium becomes saturated with the constituents of the steel. Since the constituents, i.e., iron, chromium, nickel, etc., have generally different solubility characteristics, solution attack can occur preferentially (e.g., because of a greater solubility of nickel in sodium the surfaces of austenitic stainless steel can become ferritic by the depletion of nickel). Changes in surface composition

due to solution attack by low oxygen sodium, which may be typical for a 316 SS hot leg (1200°F) pipe, are shown in Fig. 3.1. Depletion of nickel, chromium, and manganese extends ~8 microns from the surface into the material. The depth of the attack is limited by the test time and the diffusion rates of these constituents in the steel. The iron content at the surface is >90%.

Type 316 SS exposed in bimetallic loops to Na at 1200°F and 1300°F shows ferrite layer formation.¹²⁴ The layer reaches a limiting depth (for specific conditions) within 1000 hr and holds there for at least 5000 hr. Ferrite layer formation is also affected by cold work which increases the rate of formation, probably because of increased diffusion rates associated with severe lattice distortion and an increase in vacancies.

It is suggested¹²⁴ that volume change accompanying structural change causes additional structural defects which may act as nucleation sites for damage mechanisms. Grain boundary penetrations may act as notches and stress raisers promoting thermal fatigue during strain cycling. Differences in thermal expansion coefficient for ferrite and austenite (6.5×10^{-6} in./in./°F vs 10.6×10^{-6} in./in./°F) may contribute to localized structural damage.

B. Impurity Reactions

Nonmetallic sodium impurities such as oxygen, carbon, hydrogen, and nitrogen may attack steel surfaces by chemical reactions. Impurity reactions are complex and not well understood. The following mechanisms have been described for oxygen.

The high corrosion rates observed with oxygen-contaminated sodium in iron-base systems have been attributed to the formation of a ferrate compound $\text{FeO}(\text{Na}_2\text{O})_2$.⁷⁶ Another view⁷⁷ supported by special experiments,⁷⁸ which, how-

ever, have not been confirmed by other investigators, holds that oxygen exists in sodium, at least partially, in the form of negative ions, and is thus available to form FeO which then dissolves in the sodium. Whatever the mechanism, test data show that austenitic stainless steel surfaces in hot legs of oxygen-contaminated sodium systems become depleted of iron and enriched in nickel and chromium, which is the reverse of the behavior observed with low oxygen sodium (Fig. 3.1).

The mechanisms of carbon transfer between the steels and sodium are not fully understood at this time. What is known is that carbon transfer is related to differences in carbon activity in metal components and sodium with carbon transfer tending to eliminate the carbon activity gradient, that is, carbon will transfer from a high activity region to a low activity region.

The activity of carbon in sodium is quite high because of its extremely low solubility. Therefore, the sodium tends to transfer carbon from external high activity regions to external low activity regions. Since carbon activity in the selected steels increases with increasing temperature, in a monometallic system carbon transfers from hot leg to cold leg components. Also, the carbon activity in ferritic steels is higher than that in austenitic steels. Therefore, in a bimetallic system carbon transfers from 2 $\frac{1}{4}$ Cr-1 Mo to the austenitic stainless steels.

The rate of carbon transfer depends upon its diffusivity in the steels if one assumes that solid state diffusion is the rate limiting step. Carbon diffusion coefficients also increase with increasing temperature. Thus, the rate of carbon transfer from hot leg to cold leg increases with increasing hot leg temperature as is observed in monometallic systems. Estimates of carburization and decarburization of components exposed to sodium under LMFBR conditions are made in the following, which is based on papers by Kolodney⁷⁵ and Goldmann.¹⁶⁶

Decarburization

With the assumption that the diffusion of dissolved carbon in the steel is rate limiting, the depletion of carbon can be estimated from Fick's law, which reduces for a slab to:

$$\frac{C_m - C_o}{C_s - C_o} = F \left(\frac{Dt}{L^2} \right)$$

where C_m = mean carbon concentration at time t

C_o = uniform initial carbon concentration

C_s = constant surface carbon concentration

D = diffusivity of carbon, cm^2/sec

t = time, sec

L = wall thickness, cm.

The following conservative assumptions are made: carbon is dissolved up to 0.10% in 304 SS and 316 SS and up to 0.15% in $2\frac{1}{4}$ Cr-1 Mo. Diffusivities of carbon for austenitic stainless steels and $2\frac{1}{4}$ Cr-1 Mo are given in Figs. 2.3 and 2.6, as determined for austenite and alpha iron by Smith³³ and Homan,¹⁰² respectively. The sodium acts as a sink, reducing the carbon content at the surface of the steel to zero. With these assumptions and convenient tables given in Reference 159, carbon distributions and mean carbon levels can be calculated for different section thicknesses as a function of temperature and time.

Fig. 3.3 shows the calculated carbon distribution in a 1-in. thick stainless steel section with an initial carbon content of 0.10%, after exposure to hot leg sodium at 1200°F for 30 years. It will be noted that the mean carbon content is reduced to 0.056% and that approximately 60% of the section thickness retains carbon at a higher than the mean level. This example and the following pertain to sections which have only one surface exposed to sodium, as would be the case for pressure containment. For sections which are completely submerged in sodium, the data are applicable to half sections.

With the same assumptions, ratios of final mean carbon content to initial carbon content have been calculated for austenitic stainless steels and $2\frac{1}{4}$ Cr-1 Mo. The results are shown in Figs. 3.4 and 3.5, respectively. The ratios are plotted as functions of sodium temperature for three selected wall thicknesses. It is apparent that the carbon loss from $2\frac{1}{4}$ Cr-1 Mo is much more severe.

Carburization

Carburization is considered to be limited by the available sources of carbon in the system and the manner in which austenitic stainless steels are carburized. By assuming a limited supply of carbon, it is implied that there is no carbon contamination of the sodium from external sources, or if there was such contamination, the sodium purity monitors would detect it; the plant operators would reduce temperatures and purify the sodium before excessive amounts of carbon could diffuse into the component materials.

Carburization of the austenitic stainless steels proceeds via carbide precipitation, which results initially in a highly carburized case rather than a continuous carbon distribution. If it is assumed that all the carbon in a $2\frac{1}{4}$ Cr-1 Mo steel (initial carbon content, 0.12%) cold leg is transferred to an austenitic stainless steel hot leg (with equal weight of cold and hot leg piping), the carbon

case thickness will be less than 4% of the wall. Since the case thickness represents a small fraction of section thickness, and since the bulk of the material will retain normal properties, any detrimental effects of the initial carburization need be considered mainly in regard to those properties which are sensitive to surface conditions. They include fatigue, impact, and ductility.

It should be noted that if the carburized case is established early during the plant life, there will be some diffusion of the carbon into the base material, the amount depending on temperature. If the effective diffusion coefficient is as low as $\sim 10^{-11}$ cm²/sec at 1200°F and even lower at lower temperatures, as reported in References 158 and 173, most of the carbon will remain in the case during the lifetime of the plant.

Other impurity reactions are described in Section 3.2.2.

C. Mass Transfer

Solution attack and impurity reactions are treated separately in the foregoing, but the presence of one impurity in sodium alters the solubility or solution kinetics for other impurities and other materials. Thus, in practice, solution attack and impurity reactions are interdependent. Solution attack and impurity reactions, together with temperature and materials differences, frequently cause mass transfer around a sodium loop.

“Mass transfer” in the classical sense occurs because the flowing sodium tends to dissolve materials in the high-temperature regions, where their solubilities are highest, and then deposits these materials in the low-temperature regions, where their solubilities are lowest. Thus, when sodium returns to the high-temperature region, it is unsaturated and is capable of further attack. In this manner mass is transferred during each circuit. The observed mass-transfer deposits in the cold regions are usually smaller by orders of magnitude than would be predicted from equilibrium-solubility considerations. At least two factors contribute to this difference: (1) solubility rates are low in relation to loop transit times; and (2) some material precipitating in the bulk stream is not deposited on surfaces, but is swept back into the higher temperature regions where it redissolves.

As indicated above, mass transfer does not have to be solution controlled. For instance, temperature dependent oxidation-reduction reactions form a basis for mass transfer by oxidation in the hot leg and reduction in the cold leg. While mass transfer usually refers to metallic elements, the term also may be applied to the transfer of carbon or nitrogen. Carbon transfer is thought to be controlled by activity differences due to dissimilar metals or temperature differences (see Section 3.2.2).

The foregoing discussion relates to closed systems. Of course, the same mechanisms of corrosion and mass transfer apply to open systems, e.g., systems in which there are accidental contaminations of sodium due to air leaks or oil spills, or an introduction of new surfaces via refueling or repairs.

Fig. 3.2 is a representation of the mass transfer in a simulated LMFBR primary system as reported by Wozadlo.¹³⁹

D. Engineering Consequences

The engineering consequences of corrosion are depicted in the diagram in Section 3.2.1 (A).

Surface removal, normally occurring in the hot leg, leads to thinning of a pipe wall and consequent loss of load-carrying capability. In the pipe design, this phenomenon will be considered in terms of a "corrosion allowance" which necessitates increasing the wall thickness to compensate for the loss. Since metal removal rates are low, corrosion allowance is small and the penalty due to higher thermal stresses is also small.

Surface deposits, generally occurring in the cold leg, are not expected to be significant in terms of heat transfer or pressure drops, for the large size LMFBR piping. However, in a primary system, deposits of materials, which had been activated in the core regions of the reactor, could lead to special shielding requirements for the piping external to the reactor.

Changes in composition of the base material are expected to have the largest effect on the piping design, because of associated changes in mechanical properties of the selected materials. The following sections emphasize this aspect of corrosion, both because of its importance and because of a lack in the state-of-the-art.

3.2.2 Effects of Impurities

A. Oxygen

Oxygen will increase the surface removal rates in hot leg piping. The reported surface removal rates (e.g., mils/year) are usually calculated from weight losses of test specimens with the assumption that weight losses are uniform from all sodium exposed surfaces of the test specimens. The surface removal rates vary linearly with oxygen content of the sodium. While direct comparisons among various investigators are not possible due to variations in heats of materials, test conditions, and methods for measuring oxygen activities, definite decreases in attack rate are found with decreasing oxygen.

One explanation offered for increased corrosion in the presence of oxygen is the formation and reduction of the double oxide $(\text{Na}_2\text{O})_2 \cdot \text{FeO}$ in the hot and cold sections,⁷⁶ respectively, of a loop. Test data show that surface removal rates vary approximately linearly with oxygen content.^{29,68}

The oxidation behavior of stainless steel exposed to flowing sodium has been studied under different sets of conditions which may be grouped under three categories:¹²⁹

1. In sodium having less than 10 ppm oxygen;
2. In isothermal, low velocity systems, containing sodium with oxygen levels from 20 ppm to 300 ppm;
3. In nonisothermal, high velocity sodium systems with oxygen levels of 25 to 35 ppm.

A number of investigators have been concerned with the first category and have indicated that there is no retainable oxidation product on stainless steel surfaces up to 1400°F. Elemental mass transfer does occur resulting in the formation of a ferrite layer.

MSAR work has been concerned with Category 2 and is summarized in a recent report,¹²⁹ from which the following is abstracted:

At 1200°F in high oxygen sodium a surface scale forms which penetrates along grain boundaries to the base metal. The scale has been identified as sodium chromate ($\text{Na}_2\text{Cr}_2\text{O}_4$) by X-ray, electron micrography, and electron microprobe analyses. The surface scale is easily spalled off and is interspersed with ferritic iron islands. Iron and nickel concentration in the scale vary between 0.1 and 0.3%. In high velocity oxygen-contaminated sodium, whole grains can be eroded from the surface due to grain boundary attack.

At 1400°F, the scale is green and does not penetrate as deeply as at 1200°F. Relative compositions and penetrations for the two temperatures are shown below.

	<u>Na</u>	<u>Cr</u>	<u>Ni</u>	<u>Fe</u>	<u>Visible Penetration</u>
1200°F	—	46.5	0.2	0.15	0.0008 in./5687 hr
1400°F	—	36.0	0.2-0.3	1.5	4 μ /500 hr

The scale is composed of two components, a major and a minor phase. When heated to 1200°F in pure sodium, the major phase transforms to $\text{Na}_2\text{Cr}_2\text{O}_4$.

Russian workers¹³² have found that the oxide film of austenitic SS is retained in sodium up to 500 to 540°C, therefore corrosion up to 500°C depends on the extent of passive film.

Other experiments were performed by these workers using sealed tubes containing sodium with oxygen plus electrodes for passing current through the sodium at test temperatures. Oxygen was transported to anode spheres indicating transfer of negatively charged ions.

Rate of dissolution of iron increases 16-fold when oxide content of sodium increases from 0.01 w/o to 0.04 w/o.¹³²

Corrosion depth of 304 SS at 300°C in sodium for 2500 hr with 0.8 w/o O is the same as at 710°C in sodium with 0.003 w/o O for 3000 hr.¹³²

Under normal operating conditions the oxygen level in LMFBR sodium systems is expected to be 10 ppm or less, which is substantially below the levels which were used in the aforementioned tests to investigate the effects of oxygen contamination.

Metal loss rates^{29, 64, 65} showing the effect of oxygen and temperature are plotted in Figs. 3.6 and 3.7. As can be seen in Fig. 3.2, the total metal loss in the LMFBR sodium with design oxygen content of 10 ppm is about 0.6 mils per year at 1200°F for the 18-8 stainless steels. The 304 SS may be assumed to behave similarly to 316 and 18-8-Ti (321 SS).⁶⁶ Since 2¹/₄ Cr-1 Mo behaves similarly to the 18-8 stainless steels at ~25 ppm oxygen, it may be expected that its corrosion rate at ~10 ppm oxygen also will be low. Fig. 3.7 furnishes additional data on the effects of both temperature and oxygen content on the surface removal rates of 316 SS in sodium. It compares British²⁹ and American⁶⁵ data with good agreement. They show about a five-fold increase in corrosion rate at 1200°F for an increase in oxygen concentration from ~10 ppm to as much as 50 ppm. Fig. 3.8 shows that a rise in oxygen concentration from ~10 to ~27 ppm increases the weight losses of 316 SS specimens immersed in flowing sodium at 1337°F.⁶⁴ A velocity increase in the middle of the higher oxygen run had no apparent effect on the weight loss rate. A reduction in oxygen content showed the reverse effect in Fig. 3.9.²⁹

The linear relation between corrosion rate and oxygen content of the sodium at 1200°F is illustrated in Fig. 3.10. Although this figure is for 316 SS, 304 SS and 2¹/₄ Cr-1 Mo exhibit a similar relation.^{64, 66, 67} The British (UKAEA) measurements²⁹ indicate a much more rapid increase in corrosion rate than the GE data.⁶⁸

In a recent Brookhaven National Laboratory report,¹⁴¹ corrosion rates of stainless steel as determined by workers at GE, BNL, and in the UK are compared (see Fig. 3.11). Wozadlo¹³⁹ also compares most recent GE corrosion rate results with their previous results and UK results. The recent GE studies tend to corroborate the UK corrosion rate data. This is shown in Fig. 3.12.

Table 3.1 summarizes these data. It is readily seen that the corrosion and oxygen content are proportional up to 50 ppm oxygen.⁶⁶ The surface removal rates increase four times for about a four-time increase in oxygen content. However, between moderately high and very high oxygen levels (from ~50 ppm to ~300 ppm O₂) this relationship is no longer followed and the increase in surface removal rate is about tenfold for a two to six times increase in oxygen.⁷⁰

In summary, weight loss tests of austenitic stainless steels and 2 $\frac{1}{4}$ Cr-1 Mo alloy steel in sodium indicate that at 1200°F (max) the surface removal rates of the selected materials can be limited to a level of a fraction of one mil per year,^{50, 64, 67, 68, 72, 73} by controlling oxygen levels to 10 ppm (max). This assumes that surface removal rates are not strongly affected by hydraulic diameter. (See Section 3.2.3.)

B. Carbon

Carbon is an impurity that can significantly change the properties of the base materials. The transfer of carbon in sodium systems is discussed on the basis of diffusion theory in Section 3.2.1 (B). In the following, the results of actual test exposures in sodium environments are discussed. Section 3.3 covers the effects of carburization and decarburization on the mechanical properties of the candidate piping materials.

Although carburization is of great concern, there does not appear to be a widely accepted limitation for the maximum carbon content of austenitic stainless steel. In particular, the effects of long-time aging on properties have not been measured for various carbon contents. It is known that aging at temperatures in excess of 1000°F produces carbide precipitation with profound effects on ductility and toughness. For example, in 1962, Sinizer and Pearson²⁷ predicted an infinite life for 304 SS if its surface carbon content remained 0.5% or less. This prediction was based upon retention of elongations in excess of 20% at both room temperature and 1200°F after carburization in sodium, and homogenization to uniform carbon content. However, it is not clear that the homogenization period was sufficiently long to insure complete carbide precipitation. Similarly, Coobs,²⁸ at ORNL, reports that carbon concentrations up to 0.70% reduce the room-temperature elongation from 57% to 32%. However, again it is not clear that sufficient aging was performed

to insure full carbide precipitation. On the other hand, Thorley and Tyzack²⁹ carburized 18-8-Ti and 316 SS as well as variants of 316 SS and 316L SS for periods up to about nine months at temperatures of 1200°F and 1335°F. Their results are shown in Table 3.2. It will be observed that although the room-temperature and elevated-temperature ductilities are very substantially reduced, appreciable room-temperature ductility is surprisingly retained even after nine months. No impact data are given. It should be emphasized that although the exposure periods were long, carburization was occurring during the entire period and more severe deterioration might be expected from post-exposure aging at the same final carbon contents.

Despite these indications of a tolerance for relatively high carbon contents, work at MSA³⁰ has shown substantial loss of ductility in 304 SS and 316 SS with minor increases in carbon. This subject is discussed further in Section 3.2.

Loss of ductility may not be the most severe effect of carburization. Reduction of impact strength (see Section 3.2.2) may be more significant. For example, exposure of 316 SS to high carbon sodium for 4000 hr at 1200°F decreased the impact strength by 82%. About one-half of this loss may be attributed to the carburization. The effects on 304 SS are perhaps even more severe.

Although the chromium carbides are stable, decarburization of austenitic stainless steels may take place if the surrounding medium is sufficiently low in carbon activity. This low carbon activity may be produced in systems of a single alloy by transfer of carbon from hot to cold regions of the system. The transport of carbon would be accelerated if a hot-trap with carbon-gettering materials were present. Loss of carbon would result in a reduction of strength. For example, Fig. 3.13 shows the loss in rupture strength due to decarburization for a 304 SS after exposure in zirconium-gettered sodium.³¹ The loss in rupture strength for 316 SS due to similar conditions was not as great.

It should be noted that the damage to mechanical properties has been reported for relatively thin sections. It is to be expected that the detrimental effects would be less severe in thick sections. Experimental work on this factor and on the results of aging appear warranted. In addition, the transport of carbon caused by large temperature differences should be examined.

Extensive measurements of carburization rates of austenitic stainless steels in sodium have been performed in capsules at AI,³² and for periods in excess of 5000 hr in loop tests at MSA.^{100 102} The most reliable tests appear to be those employing saturated sodium; these are described below. The

results of these measurements have been used to deduce coefficients for the diffusion of carbon in carburized stainless steel. However, although the diffusivities calculated by both organizations are in good agreement, they are not true values because the calculations were based on the total carbon content of the surfaces (including precipitated carbides) rather than dissolved carbon. As indicated in Section 2.3.1(C), Kolodney¹¹⁹ has proposed a mechanism that satisfactorily explains the observed carburization rates of stainless steels.

Carburization of 304 SS and 316 SS

Fig. 3.14 provides information on the rates of carburization of 304 SS at various temperatures in high-carbon sodium.^{50,70} Over 10,000 hr (over 1 yr) at 1200°F is required for a carburization depth of 40 mils. As is noted later, decarburization occurs more rapidly. The mechanism of carbon penetration comprises a first-stage increase in grain boundary carbide followed by general intragranular carbide precipitation. Most of the carbon is present in carburized stainless steel as an insoluble carbide rather than as dissolved carbon.^{32,50,68,70,74,75}

The average surface concentration of carbon in a carburized stainless steel case after prolonged exposure to sodium is reported by Al¹⁵⁸ to be ~3.2% at 1200°F.

Carbon transfer was studied¹²⁴ in isothermal pumped sodium loops. Tubular samples were exposed for up to 700 hr at 1050°F. Types 316 L, 316, and 304 peaked at ~350 hr in C pickup with maximum C between 0.09 and 0.08. Type 347 showed a slower rate of pickup with no peak up to 700 hr.

Decarburization of 304 SS and 316 SS

See Section 2.3.1 (E).

Decarburization of 2¹/₄ Cr-1 Mo Steel

The 2¹/₄ Cr-1 Mo steel can lose carbon to austenitic stainless steels. With sufficient carbon transfer, the stainless steel is embrittled and the 2¹/₄ Cr-1 Mo is weakened. The 2¹/₄ Cr-1 Mo also can be depleted of carbon in the hot leg of a monometallic system by carbon transfer to the cold leg.

The rate of decarburization is temperature and oxygen dependent, as shown in Figs. 3.15, 3.16, and 3.17. These data were obtained in a monometallic loop of 2¹/₄ Cr-1 Mo. The effect of temperature is predominant

with the decarburization becoming appreciable in the vicinity of 900 to 1000°F, as shown in Figs. 3.15 and 3.16. Fig. 3.16 shows the effect of exposure time (to +19,000 hr) and temperature on decarburization. Decarburization of the 1100°F specimens is essentially complete after only 2500 hr. Carbon content in the sodium was not given. A much thicker $2\frac{1}{4}$ Cr-1 Mo pipe section in the same loop as the specimens in Fig. 3.16 was "completely" decarburized (i.e., to 0.025 w/o C) at the end of the test (~19,000 hr), although exposed to the sodium from one side only. The rate of carbon loss is not affected significantly by oxygen content in the levels of 10 to 50 ppm in the temperature range of 700 to 1100°F (Fig. 3.15). Where gross increases in the oxygen content of sodium occur, at levels of 30 and 300 ppm, decarburization increases significantly at the higher oxygen level, Fig. 3.17.^{79,80}

The additions of strong carbide-forming elements, Cb, Ti, and V, to chromium-molybdenum steels have been successful in helping to prevent decarburization by sodium.⁸¹⁻⁸³ It has been found that the most effective carbon stabilizer elements, columbium and titanium, do not produce the best steel.⁸⁴ Impact strength and weldability are poorer for columbium-stabilized $2\frac{1}{4}$ Cr-1 Mo than for the unmodified alloy. The addition of a small amount of nickel to the Cb-modified alloy significantly improves impact values but not to the level of the unmodified or V-modified alloys. Although vanadium was found to be a much less effective carbide stabilizer than Cb and Ti, additions of 0.6 w/o V resulted in an alloy of far better toughness and weldability. Three 0.6 w/o V additive alloys of $2\frac{1}{4}$ Cr-1 Mo were compared with unmodified $2\frac{1}{4}$ Cr-1 Mo alloys in carbon activity. Two of the three V-modified $2\frac{1}{4}$ Cr-1 Mo alloys also contained small additions of columbium. The results in Fig. 3.18 showed that a vanadium-modified $2\frac{1}{4}$ Cr-1 Mo steel had a significantly reduced decarburization rate in sodium. The next best alloy in mechanical and welding properties, the $2\frac{1}{4}$ Cr-1 Mo + 0.6 V + 0.1 Cb, showed only a slight improvement in preventing decarburization. None of these alloys are now commercially available in the U.S. and they are not code materials. Some of these modified low-chromium alloys are widely used in Europe.

More than doubling the Cb (to 0.25 w/o) produced a marked improvement in carbon stability, but at the sacrifice of impact strength and weldability. Additions to $2\frac{1}{4}$ Cr-1 Mo of 0.73 w/o Cb, 0.4 Cb + 0.4 Ti, resulted in alloys with complete resistance to decarburization but with even greater mechanical and welding property disadvantages than above.⁸⁴ The addition of various amounts of chromium (to 12 w/o), a fairly strong carbide former, did not provide expected decarburization suppression to the extent provided by much smaller additions of Cb, Ti, and V.^{82,85} Higher chromium alloys (9 Cr-1 Mo) showed high carbon losses in 1300°F sodium containing 100 ppm oxygen.⁸²

In more recent work¹²⁸ the basic alloy (10 CrMoNiNb 910) – Thyssen designation (3 HK 53) – has been evaluated by interaction in sodium for up to 3000 hr at 580°C. Complete stabilization of carbon and nitrogen was achieved with TaNb/C + N of 8.7 (the alloy specification requires 10:1 min). MSAR investigations¹³⁰ confirmed the superior stability of this alloy for up to 4000 hr at 650°C in static sodium. Bauer and Conner¹²³ have summarized the recent progress in the stabilization of 2 $\frac{1}{4}$ Cr-1 Mo steel. From their survey the following information is of interest:

The Japanese¹⁸⁵ found that addition of Nb to 2 $\frac{1}{4}$ Cr-1 Mo prevented decarburization at 600 and 700°C in diffusion couples. These materials also showed improved stress-rupture life at 550°C compared with unstabilized.

In a Russian study¹⁸⁶ it was observed that the rate of decarburization increased with stress.

Czech work¹⁸⁷ indicated that decarburization of ferritic steels was independent of oxygen content in sodium above 3 ppm which confirms British¹⁸⁸ work. However, the Russians¹⁸⁹ found a dependence.

ANL work¹⁸³ predicts from thermodynamics the carbon transfer in systems involving sodium and austenitic and ferritic steels. Thermodynamic activity and solubility of carbon in Fe-Cr-Ni alloys were calculated for a temperature range of 500 to 800°C by extrapolation of higher temperature data for Fe-Cr-C and Fe-Ni-C systems. Results show that ferritic alloys with varying C may carburize or decarburize depending on C concentration in Na and temperature. Also shown are the effects of Cr depletion by solution or precipitation on carburization potential. Both decrease it.

Additional details of the MSAR work^{130,136} are summarized below:

Six ferritic steels were tested in static pots to determine the extent of decarburization in 1200°F sodium. Four were based on 2 $\frac{1}{4}$ Cr-1 Mo and two on 5 Cr-1 Mo. Three of the 2 $\frac{1}{4}$ Cr-1 Mo steels were modified to stabilize carbon. Stabilizing elements were: V or Ti or Nb/Ta for the 2 $\frac{1}{4}$ Cr and Ti for the 5 Cr alloy.

Both Ti and Nb/Ta additions to 2 $\frac{1}{4}$ Cr-1 Mo are effective stabilizers of this steel with respect to carbon transport in sodium. Ti-stabilized 2 $\frac{1}{4}$ Cr-1 Mo requires extensive post-weld heat treatment to achieve weld ductility. B&W has removed this material from the market because of this problem.

Insertion of ferritic samples in test retorts resulted in an indication of increased oxygen content of the sodium despite cold trapping at <250°F. Insertion of austenitic specimens did not cause similar effects.

The effect of thermal aging for all alloys tested was to decrease both tensile and yield strength with time, while ductility was practically unchanged. All alloys showed coalescence of carbides but no grain growth. Sodium exposure resulted in loss in strength due to both thermal aging and sodium effects (generally loss of carbon). The major effect is due to carbon loss, therefore carbon variance with exposure time reflects mechanical property changes due to sodium exposure. Figs. 3.19 through 3.22 summarize this.

Titanium additions to $2\frac{1}{4}$ Cr-1 Mo provide stabilization of carbon but result in poor weld ductility. The Thyssen alloy with Cb/Ta additions showed anomalous behavior under MSAR test conditions which included 1200°F exposure and a large γ/α ratio. Euratom tests involve either γ/α of 1:1 or 1:>1. Behavior at MSAR involved accelerated grain growth of surface exposed to sodium.

These studies of ferritic alloys show that exposure of $2\frac{1}{4}$ Cr-1 Mo for 10,000 hr to clean sodium (static) in a 316 SS retort at 1050°F is equivalent to 1000 hr exposure to 1200°F under the same conditions from a standpoint of carbon loss and deterioration of tensile properties.

C. Nitrogen

Nitrogen transfers in sodium systems, but little is known about the basic mechanisms involved. Severe nitriding of steel surfaces has been observed at interfaces between liquid sodium and nitrogen, where nitrogen was used as a cover gas.⁸⁶⁻⁸⁸ LMFBR cover gas will be helium or argon to avoid this problem.

However, it has been noted in operating reactors (LAMPRE-I, EBR-II, and EFAPP) that nitrogen levels in the helium or argon cover gases gradually increase after sodium circulation has started. LASL has concluded that nitrogen was entering the LAMPRE-I system by diffusion through all parts of the coolant loops.⁸⁹ They purged the cover gas whenever the nitrogen level reached ~600 ppm, but state that "whether this was a useful or an unnecessary precaution has not been determined."

In the past, denitriding has not been considered to be a problem. Changes in nitrogen levels generally have not been measured. At temperatures of 1200°F or less, they would be expected to be small for the normal test times of a few thousand hours. However, denitriding might become significant in a pipe which is completely submerged in a sodium pool for periods of over 200,000 hr.

There are some data at higher temperatures which suggest that de-nitriding might have to be considered for completely submerged pipes. For example, in a 316 stainless steel, Nb-1% Zr system in which 0.060-in. thick 316 SS specimens were completely submerged in potassium at 1600°F, their nitrogen content decreased from initial values of ~500 ppm to less than 50 ppm in 5000 hr. The loss of nitrogen in the adjacent pipe wall was negligible, presumably due to the availability of nitrogen from the surrounding air.⁹⁰

Chromium nitride was found on surfaces of austenitic SS (similar to 321) when nitrogen was used as a cover gas with sodium at 450°C in an isothermal system operated for 1 yr. Elongation prior to exposure was 44%; after exposure it was 34%.¹³²

D. Hydrogen

Hydrogen at the levels encountered in the piping system should not affect either the stainless steels or the 2 $\frac{1}{4}$ Cr-1 Mo. Hydrogen could affect the activity of oxygen and carbon in the sodium by forming compounds. As with nitrogen, hydrogen levels in cover gases have been observed to increase with time. The sources of the hydrogen have not been definitely identified.

3.2.3 Effect of Velocity – Reynolds Number

If diffusion rates through the sodium boundary layer represent the rate-controlling step for the corrosion of pipe walls, then one would expect surface removal rates to be directly related to velocity and hydraulic diameter, or Reynolds number. For such a situation, Epstein derived an equation which shows that the surface removal rates vary with velocity to the 0.8 power and hydraulic diameter to the -0.2 power.⁹¹ He also showed that for solution-rate-limited corrosion, surface removal rates are independent of geometry and velocity, as might be expected.

In all of the literature on sodium corrosion test results, hydraulic diameter is not considered to be a variable (except for L/D ratios in the downstream effect - see Section 3.2.6). This omission is excusable because all weight loss tests have been conducted with passages of relatively small hydraulic diameters. Also, because of the expected minor dependency, the effects of any small changes in hydraulic diameter would be difficult to detect. Yet, this leaves a problem for the prediction of surface removal rates in a 48-in. pipe. This dimension is 100 to 1000 times larger than the hydraulic diameters for which corrosion test data are available. If the rates vary as the -0.2 power, surface removal rates in a 48-in. pipe would be correspondingly 40% to 25% of those reported in the literature. It would seem desirable to generate test data to confirm such predictions. This will be considered as

a subject for future R&D. The following discussion is in regard to velocity effects.

Test results on velocity effects are contradictory. Data obtained in six loops at General Electric Co. confirm Epstein's prediction for the boundary layer diffusion limited case. For these particular loops and tests it was possible to correlate the data with the following empirical equation:⁹²

$$R = V^{0.884} O_X^{1.156} \exp \left[12.845 - \frac{23,827}{T + 460} - 0.00676 (L/D_i) + \frac{2.26}{t + 1} \right]$$

where R = steady-state rate of metal removal, mg/dm²-mo

O_X = oxygen concentration in the sodium, ppm

V = sodium velocity, ft/sec

T = corrosion site temperature, °F

L/D_i = downstream factor (site location in number of hydraulic diameters downstream from beginning of isothermal section)

t = exposure time, mo.

The values of the constants and exponents in this equation were established by applying statistical, best-fit methods to the test data. It will be noted that the velocity exponent is 0.884, close to the 0.8 prediction. This exponent was applicable to the austenitic stainless steel and low chromium alloy steel data.

Continued mass transfer studies at GE^{113, 139, 167-170} have resulted in a modification of the overall corrosion rate prediction equation for the hot leg for 316 SS and 304 SS. The modified equation is given below.

$$\ln CR = 14.55 + 0.435 \ln (V) - 0.000958 (L/D)_1 - 0.122 \ln (L/D)_2 \\ + 1.445 (1/t + 1) + 1.156 \ln O_2 - \left[\frac{23,827}{T + 460} \right]$$

where CR = rate of weight loss to time t, mg/dm²-mo

V = sodium velocity, ft/sec

T = sample temperature, °F

O₂ = oxygen content of the sodium, ppm

t = months exposure

(L/D)₁ = downstream factor = number of hydraulic diameters downstream in an isothermal region

(L/D)₂ = downstream factor = number of hydraulic diameters downstream in an iso-velocity region.

The earlier (Phase I) work was concerned with temperature, velocity, oxygen level, and downstream distance as the primary variables. In the Phase II work, the effects of oxygen and velocity were not studied. A detailed discussion of the comparisons between the two equations is given in Reference 139.

However, there are test data obtained in other systems, which do not corroborate these results. Fig. 3.8, which was discussed previously, shows no change in surface removal rates when the velocity was changed from 26 ft/sec to 10 ft/sec at ~10 ppm oxygen or from 10 ft/sec to 16 ft/sec at 20 to 30 ppm oxygen.

The presentation of selected data in Fig. 3.23 suggests a velocity-oxygen interrelation. However, the 30 to 50 ppm oxygen line detracts from the conclusion which might otherwise be drawn, i.e., at low oxygen levels the surface removal rates are independent of velocity. For intermediate oxygen levels, velocity dependence increases with oxygen content up to some limiting velocity (e.g., 10 ft/sec at 25 ppm).

Wozadlo and Spalaris¹³⁹ have summarized GE and UK corrosion work and have proposed that at sodium velocities between 0 and 10 ft/sec, diffusion into the liquid is the rate controlling step, while at velocities over 15 ft/sec, either solid-state diffusion or reaction kinetics is rate-determining. Therefore, a single velocity coefficient cannot adequately describe the velocity dependence over a wide range of velocities.

Because the removal rates of iron and chromium from stainless steel in contact with sodium are dependent upon both oxygen level and temperature (see Fig. 3.24), there is a concomitant velocity dependence of corrosion rate at high oxygen and temperature levels. The velocity effect is shown in Figs. 3.25, 3.26, and 3.27, with recent GE results now in agreement with UK.

Some R&D is indicated to resolve these differences and will be defined in a subsequent task.

3.2.4 Effects of Temperature and Temperature Difference

As already noted, surface removal rates and mass transfer are strongly dependent on temperature and temperature differences (i.e., hot and cold leg effects). This section contains a brief review of temperature relationships and the effects of temperature difference.

A. Temperature

Surface removal rates vary exponentially with temperature. They increase by a factor of 2 to 3 for each 90°F temperature rise in the 1100 to 1300°F temperature range at oxygen levels above 15 ppm for ferritic steels and above 5 ppm for austenitic steels.^{29,65,74}

The dependence of surface removal rates on temperature with both oxygen content and velocity held constant follows the Arrhenius relationship given below and also illustrated in Fig. 3.6 for the temperature range 830 to 1300°F.²⁹

$$R_c = 2.3 \times 10^{-6} \exp^* \left[\frac{-17,500}{RT} \right] \text{ cm-sec}^{-1}$$

where R_c = the surface removal rate

T = temperature, °K

R = the gas constant.

Temperature influence on surface removal rates is independent of the oxygen concentration, up to about 50 ppm oxygen.⁶⁶ Temperature effects on surface removal rates in sodium are about the same for austenitic and ferritic steels.^{65,93} Activation energies of about 18 to 25 K-cal are reported for austenitic steels, and ferritic steels have approximately the same values.⁶⁶ An increase in maximum system temperature will also increase the mass transfer deposition rate. The effect on cold leg deposition with varying maximum temperature for a constant ΔT of 300°F is shown in Fig. 3.28.

B. Temperature Difference, ΔT

An increase in mass transfer rates results from an increase in the temperature difference in the system, i.e., by lowering the cold leg temperature and maintaining a constant maximum hot leg temperature. Metallic component solubility increases with temperature. The temperature-gradient mass transfer therefore increases, as the driving force is the difference in solubility of the dissolved metal at the high and low temperatures existing in the system.

In deriving the previously mentioned velocity dependency, Epstein also showed that the surface removal rates should vary linearly with system ΔT for either diffusion or solution limited corrosion.⁹¹

*Means e[].

Few data are available on surface removal rates as a function of system ΔT .⁷⁴ GE⁶⁷ assessed the effect of two values of ΔT , 250 and 500°F, on 316 SS and 2 $\frac{1}{4}$ Cr-1 Mo in monometallic loops. These loops contained flowing sodium at 1100 or 1200°F maximum temperatures and 12 to 50 ppm oxygen levels. They concluded that the data did not show a statistically valid trend, and therefore, omitted any ΔT term from the surface removal rate equation (Section 3.2.3).

However, they did find statistically significant differences in cold leg deposition rates for different ΔT 's. Thus, the following equation for cold leg deposition rates does include ΔT in the exponential.

$$R = O_X^{0.94} \exp \left[(22.61 - 0.038 \Delta T) - \frac{(31,980 - 55.24 \Delta T)}{T + 460} - 0.0032 (L/D_i) + \frac{0.92}{t+1} \right]$$

where R = steady-state deposition rate of metal, mg/dm²-mo

ΔT = maximum temperature difference, °F

T = cold leg temperature (at deposition site), °F

L/D_i = downstream factor (site location in number of hydraulic diameters downstream from beginning of isothermal section)

t = exposure time, mo.

While, according to this equation, an increase in ΔT will increase the rate of metal deposition, the relation is not linear.

Fig. 3.29 illustrates the effect of ΔT for Inconel. These data are presented because of a lack of extensive experimental data on the effect of ΔT for the primary materials at the temperatures of interest. For the conditions of Fig. 3.29, the effect of ΔT on deposition rate is linear. Fig. 3.28, although a plot of deposition rate vs system temperature, shows the effect of a specific ΔT (300°F) on cold leg deposition rates of several alloys. It indicates that for the conditions described, 304 SS and 316 SS are far less subject to cold leg deposition than the higher nickel content Inconel.

3.2.5 Effect of Time

Weight loss rates of the primary materials in sodium were found to be essentially constant with time after several hundred hours of operation.^{64, 94} The straight lines in Figs. 3.8, 3.9, 3.30, and 3.31, which are plots of weight change vs time for 321 SS and 2 $\frac{1}{4}$ Cr-1 Mo and 316 SS, illustrate this. During

an initial transient period when this straight-line relationship does not hold, higher positive and negative (weight loss and weight gain) rates have been observed. The large differences shown for initial rates have not been fully explained, but might be attributed to different test conditions and equipment.

Information on the depletion of chromium in 316 SS upon exposure in low-oxygen sodium are shown in Fig. 3.32 for hot leg samples. The corresponding chromium-rich deposits on $2\frac{1}{4}$ Cr-1 Mo surfaces of cold leg samples with time are shown in Fig. 3.33. As expected, the chromium depletion is both time and temperature dependent.

The chromium-rich deposits build up with time and appear to reach an equilibrium composition after 5000 hr (Fig. 3.33). Nickel depletion from austenitic steels and deposits on cold leg surfaces will follow a similar trend curve vs time. The depletion relationship between Cr and Ni is shown in Fig. 3.1, discussed in Section 3.2.1.

In summary, the surface removal rates of the selected materials are high initially, decrease with time, and then level off to a constant rate, if the environment remains constant. Where oxygen impurity levels decrease with time, the corrosion rate will keep dropping. Since carbon activity gradients usually decrease with time, carburization and decarburization rates also will decrease. Extreme care must be exercised in any attempt to extrapolate relatively short-time experimental corrosion rates to anticipated longer-time system service periods, since corrosion and mass transfer rates might change as impurities in the system vary with time.

3.2.6 Downstream Effect

In long-time corrosion and mass transfer tests performed at GE, it has been found that the corrosion rate is dependent upon position even when the sodium temperature and velocity are constant. The corrosion rate decreases in the direction of flow, i.e., downstream.^{29,68,95} Similarly, when metal is being deposited in a cold region, the rate of deposition is progressively reduced downstream. To account for this phenomenon in the statistically derived equations given in Sections 3.2.3 and 3.2.4, a factor L/D_i has been inserted in the equations. This factor defines position in terms of the number of hydraulic diameters downstream from the beginning of an isothermal section.⁶⁷

An obvious explanation attributes the effect to the increase in concentration of dissolved metals in sodium in the direction of flow in the hot zone and a corresponding decrease in the cold zone. The driving forces for solution or deposition would thereby be reduced in the flow direction. Unfortunately,

this simple explanation is difficult to sustain in view of measured solubilities and deposition rates of the dissolving metals. Therefore, at present there is no widely accepted theoretical explanation for the downstream effect. Instead, the phenomenon is handled empirically by measurement of the appropriate L/D_i coefficients.

3.2.7 Effect of Stress

No correlation of stress with corrosion in sodium has been shown for austenitic stainless steels at oxygen levels less than 50 ppm.⁵⁰ Zebroski et al.,⁶⁸ however, express the opinion that stress might influence increased intergranular attack on stainless steel (i.e., 316 SS). They state: "Relatively low macrostresses may induce locally high microstresses to initiate and propagate intergranular attack." The $2\frac{1}{4}$ Cr-1 Mo is subject to minor intergranular attack at moderately high oxygen levels (30 to 50 ppm), if subject to localized stresses equal to or above 50% of the yield stress.⁹⁶

Stress is also reported to enhance carbon diffusion in $2\frac{1}{4}$ Cr-1 Mo,⁶⁶ increasing carburization rates, but no such effect has been observed with respect to the carburization rates of 304 and 316 SS.

3.2.8 Local Corrosion (Intergranular and Pitting Attack)

Generally, sodium is not expected to cause intergranular and pitting attack of serious consequence for the selected piping materials. Intergranular attack has been found in $2\frac{1}{4}$ Cr-1 Mo piping in both bimetallic and all-ferritic loops.⁹⁶ The greatest depths of attack (2 to 3 mils) occurred in the highest temperature portions (1200°F) of the loops after 20,000 to 28,000 hr exposure. In the lower temperature regions of 900 to 1000°F, the attack amounted only to 1/2 mil in the form of pits and surface etching. No intergranular attack on $2\frac{1}{4}$ Cr-1 Mo was observed at moderate oxygen levels of 10 to 15 ppm, and it was thought necessary for oxygen levels to equal or exceed 30 ppm to produce attack in the presence of low stresses. Only 5 mils intergranular attack on $2\frac{1}{4}$ Cr-1 Mo were allowed for in 30 yr.⁹⁶

Intergranular attack was found in four "cross-over" pipes of the GE loops which had 316 SS hot legs.⁶⁷ The cross-over pipe is an isothermal section which connects the heating leg with the cooling leg of these loops. The penetration amounted to 7 mils after 30,000 hr ($\sim 3\frac{1}{2}$ yr).

This attack has been further investigated by GE⁶⁸ and tentatively attributed to a combination of sensitization of the 316 SS at $\sim 1200^\circ\text{F}$, thermal stresses occurring during startup which can initiate and propagate the intergranular attack and a chemical attack due to a concentration of sodium oxide and possibly

sodium hydroxide. The presence of oxide/hydroxide can be explained by the introduction of air and moisture during sample introduction to the loop with subsequent formation of a film in the cross-over leg because of its high position in the system.

Attack was postulated to occur only shortly after startups and not during normal long-time operation. Further data are being obtained by GE to establish more thoroughly the causes for the attack and methods to prevent it.

Examination of material showing attack revealed that there was severe sensitization in the form of continuous sheets of minute angular $M_{23}C_6$ particles and that corrosive environment attacked certain crystallographic planes preferentially. Mechanisms for this attack are associated with intermittent atmospheric contamination (forming $NaOH + Na_2O$) and with large thermal stresses during startup.¹²⁴

Later GE work¹³⁹ has indicated that localized attack is possible if oxygen level in sodium is >20 ppm. For example, plain carbon, low-alloy, ferritic stainless and austenitic stainless steels were exposed at $1175^\circ F$ in static sodium (100 ppm oxygen) for 596 hr.¹⁴⁰ Intergranular attack was found and was similar to that previously reported in cross-over piping.

Intergranular attack was observed only when the oxygen level of sodium was over 50 ppm in Russian work.¹³²

3.2.9 Effect of Alloying Content

Small changes in alloying content within specification limits will not affect corrosion or mass transfer rates for the primary materials.²⁹ However, according to GE studies,¹⁴⁰ the presence of Cr in solution in sodium lowers carbon activity in sodium which increases the rate of decarburization of steels. Increasing Cr in steel with oxygen in sodium results in a change from intergranular attack to general attack. At low Cr levels, intergranular attack by oxidation and decarburization in high-oxygen sodium is the result.

The interrelationship between carbon transport and chromium is shown in Fig. 3.34.¹³⁹

3.2.10 Effect of Sodium Mass Flow Rate

GE reports^{139, 140} that sodium mass flow rate has little effect on corrosion of austenitic stainless steels at velocity of 3 fps and Reynold's numbers of 8110 and 7670 for flows of 0.5 gpm and 1.0 gpm.

3.2.11 Effect of Heat Treatment – GE¹⁴⁰

Heat treatment (1650°F – 24 hr) to agglomerate carbides prior to testing resulted in no improvement in 304 SS and an apparent improvement in 316 SS. In 1300°F flowing sodium ($\Delta T = 425^\circ\text{F}$) and at 12,000 psi initial hoop stress, annealed material had a life of 830 hr and heat treated material ~1570 hr (avg two tests).¹⁴⁰

3.3 EFFECTS OF SODIUM ON PROPERTIES OF SELECTED MATERIALS

A state-of-the-art survey of the effects of sodium on mechanical properties is presented in this section and will serve as a basis for selecting allowable design properties in a subsequent task (Task 315). In the past, allowable design stresses for sodium piping have been based on the generally used materials properties which were obtained in air or inert gas environments. A comparison of mechanical property data presented below for sodium, air, and inert gas environments shows that the past practice of relying on data taken in air or inert gas can be misleading. However, as mentioned in the introduction to Section 3, the sodium environment data generated to date are inadequate for developing allowable design properties of the same reliability as those developed for air. Again, it will be obvious from the data below that, for materials in sodium, some correction must be made of air data, even if based on an inadequate amount of information. This survey will also serve as a basis for defining R&D requirements, which is to be accomplished in a subsequent task (Task 314).

Most of the available data on the effect of sodium environment have been generated by MSAR under AEC sponsorship. The materials program at MSAR was undertaken to determine whether an environment of sodium and its normally anticipated contaminants affect the physical properties of structural steels of interest for the construction of sodium system components. This objective is relatively limited. It was not the intent of the initial program to generate sufficient data to establish new code allowable design values.

Results of the MSAR Program are presented in Progress Reports,¹³⁴⁻¹³⁶ Topical Reports,^{80, 100-102, 129, 130} and Conference Papers.^{30, 133} To date, no drastic effects of exposure to clean sodium (<30 ppm oxygen, low carbon) on the mechanical properties of 304 SS or 316 SS at 1200°F have been observed. Specific conclusions and limitations may be stated as follows.

1. Low-Oxygen Exposures (1200°F)

No significant effect on the short-time tensile properties or the creep-rupture strength of 316 SS was observed due to pre-exposure to low-oxygen (<30 ppm) sodium for 4000 hr. The

fatigue life of 316 SS under strain cycling was roughly three times longer in low-oxygen sodium than in air.⁸⁰

Impact strength at room temperature decreased with time, most notably with 316 SS, as a result of exposure to low-oxygen sodium primarily due to the formation of sigma. This effect also occurred as a result of exposure to helium. No comparable exposures and tests were conducted on 304 SS.

The tensile, creep and rupture strengths of $2\frac{1}{4}$ Cr-1 Mo were reduced due to 4000-hr exposure in low-oxygen sodium. Elongation and toughness were correspondingly increased. There was little difference in the effect of oxygen level on toughness and creep rates, but high-oxygen content caused greater changes in all other mechanical properties.

2. High-Oxygen Sodium Exposures (1200°F)

The short-time tensile and creep-rupture properties of 304 SS were only slightly decreased due to exposure to high-oxygen (~300 ppm) sodium for 4000 hr. No significant effect was observed on the creep-rupture strength of 316 SS. However, after the long-time creep-rupture tests in high oxygen sodium, the ductility of 316 SS was significantly lower than when exposed in air. Fatigue data for 304 and 316 SS were inconclusive, but indicate that the environment had an adverse effect on the fatigue life measured at low cyclic strains (long exposures).

Both 304 SS and 316 SS after 4000-hr exposure in high-oxygen sodium showed considerably lower impact strength than resulted from exposure in air. It was also somewhat greater than in low-oxygen sodium. The toughness of $2\frac{1}{4}$ Cr-1 Mo was improved by exposure to high-oxygen sodium.⁸⁰

3. High-Carbon Sodium (1200°F)

Exposure for 4000 hr to high-carbon (~300 ppm) sodium at 1200°F resulted in increases in yield and tensile strengths of approximately 20% and 50%, respectively, for both 304 SS and 316 SS. Ductility in these tests was reduced 80% for the 304 SS and 96% for the 316 SS. These comparisons are with room-temperature properties. Their creep-rupture strengths were relatively unaffected except that elongations at failure were reduced by 50 to 75%. Prolonged exposure to high-carbon

sodium significantly reduced fatigue life of both alloys vs controls based on data obtained from strain-cycling fatigue tests. The impact strengths of 304 SS and 316 SS were severely decreased by exposure to high-carbon sodium compared with exposure to high-oxygen sodium. Losses were greater for 316 SS. The $2\frac{1}{4}$ Cr-1 Mo suffered only a small loss in toughness after long exposure.

The summary conclusions enumerated above and most of the following details concerning the effects of sodium environment on mechanical properties are based on the MSAR work which has the following weaknesses in application to LMFBR piping:

1. No measurements or studies were made on the effects of ΔT , velocity, and hydraulic diameter. These are expected to be important factors in sodium environmental effects on mechanical properties of steels.
2. All specimens were 1/16-in. maximum thickness. It is difficult to extrapolate significant environmental effects on mechanical properties to thicker sections of LMFBR piping. This is particularly true for carburization and decarburization effects.
3. It is questionable whether the control and measurement of oxygen levels in the sodium by the techniques used have validity.
4. It is questionable whether the designation "high-carbon sodium" has any definable relationship to carburization potential. MSA obtained so-called "high-carbon sodium" by maintaining an excess of graphite in the sodium environment.

These limitations are not expected to change the major conclusion which can be drawn, namely, sodium will not cause an unacceptable deterioration of properties. However, considerably more data are required to firm up values for allowable design properties. Such data must be obtained under conditions which are more representative of the LMFBR environment and for which more sophisticated techniques for sodium purity control are applied.

3.3.1 Effects of Sodium on Tensile Properties

A. Effect of Sodium with Oxygen Impurity on 316 SS (MSAR)⁸⁰

Data on 316 SS in Table 3.3 show that exposure to 1200°F sodium with "low" oxygen levels (10 to 30 ppm) somewhat increases the tensile strength and decreases the yield strength and elongation as measured at room temperature.

The data¹³³ in Fig. 3.35 indicate that prior exposure of 316 SS specimens for 4000 hr to 1200°F sodium (purity <10 ppm oxygen) results in insignificant changes in tensile properties when compared with unexposed specimens tested in air or helium at 1200°F, and that heat-to-heat variations are much greater than those due to the test environments. However, exposure to "high" oxygen sodium (>300 ppm) decreases all the tensile properties at room temperature, but not to below allowable design stress levels or to an unacceptably brittle condition. Elevated temperature tests at 1200°F after the same exposures show similar effects of oxygen on tensile properties, with tensile and yield strength less, and change in elongation much greater, ~50% reduction.

B. Effect of Sodium with Carbon Impurity on 316 SS (MSAR)⁶⁰

Data on 316 SS in Table 3.3 show that in tests at room temperature, previous exposure to 1200°F sodium with "high" carbon content decreases tensile strength but increases yield strength. Elongation is decreased to nearly zero. Tests at 1200°F after the same exposure show increases in both tensile and yield strengths and a similar decrease in elongation to nearly zero.

The above tensile and yield strength data are plotted vs temperature in Figs. 3.36 and 3.37 and are compared with United States Steel average values. The 25% allowable ASME Code values for 316 SS tensile strength vs temperature are plotted in Fig. 3.38 and the 62½% allowable ASME Code values for yield strength in Fig. 3.39. MSAR tensile and yield data from Table 3.3, after the same corresponding percentage modifications, are compared against code values in these respective figures. In both cases, it is apparent that the contaminated sodium exposures did not degrade either the tensile or yield strength values below the code allowable stresses. However, a plot of the above elongation data in Fig. 3.40, when compared with typical elongation data reported by United States Steel, shows how significantly elongation is reduced by exposure to high-oxygen sodium and especially carbon-contaminated sodium.

More recent MSAR data¹⁰³ on carburization in high-carbon sodium of 5-mil samples of 304 and 316 SS show complete loss of ductility when the samples reached carbon contents of about 0.6 w/o. Substantial ductility, 40% of original, was retained at a carbon level of 0.1 w/o in 304 SS. Oxygen and carbon activities in the sodium were not given. However, carbon concentration of sodium as measured in a chemical analysis is not the ideal way to characterize carburization or decarburization potential in a sodium system. The problem of defining the parameters governing carburization has been summarized by Kolodney.³⁵ It is two-fold: (1) how much carbon and what carbon distribution can be allowed in a steel before it becomes unfit for service? (2) how high a carbon activity is tolerable in the sodium without causing excess carburization over either short or long exposure times? Satisfactory

answers to the above two questions are as yet not available. Divergent opinions on the allowable carbon levels exist and have been covered in Section 2.3.1 (E).

C. Effect of Sodium with Carbon Impurity on 304 SS (SRE)

Sinizer and Pearson²⁷ studied carburization of 304 SS using a different approach than MSAR. They used carbon-contaminated sodium of unspecified carbon activity at 1200°F to carburize specimens of 304 SS to different levels. They then "homogenized by diffusion" the case-carburized specimen so that carbon content was uniform throughout. These specimens were then subjected to tensile tests. The yield strength and elongation resulting from these tests are plotted vs carbon content in Figs. 3.41 and 3.42 for room temperature and 1200°F tests. As expected, the yield strength at both temperatures increases with increasing carbon and the ductility decreases. Ductility decreases more with carbon content at room temperature than at 1200°F.

D. Effect of Sodium with Oxygen Impurity on 304 SS (MSAR)¹⁰⁰

Fig. 3.43 summarizes the tensile property results of a test program of 304 SS exposed to high-oxygen (~300 ppm) contaminated sodium at 1200°F. Tests of the exposed material were conducted at room temperature and 1200°F. The "UW" (unwashed) designation of the 1200°F test temperature is explained at the bottom of the figure, and shows the effect on tensile properties of the oxidation in air of the residual sodium on the surface of the specimen. The room temperature tensile tests in air showed a 9% reduction in tensile strength, an 18% reduction in yield strength, and a slight loss of ductility due to 1200°F exposure to "Na Hi O₂" as compared to the as-received material. The 1200°F tests of pre-exposed specimens showed approximately a 10% reduction in tensile strength, but no significant change in yield strength as compared to the as-received material. The ductility of the pre-exposed specimens dropped 23% vs the controls.

E. Effect of Sodium with Carbon Impurity on 304 SS (MSAR)¹⁰⁰

Fig. 3.44 summarizes the tensile property results of a test program of 304 SS exposed to high-carbon sodium at 1200°F. High-carbon sodium was not characterized in terms of carbon activity. Because of the extreme embrittlement of the material, which led to premature failure at the pinhole made for extensometer attachment, no tensile strength could be obtained at room temperature. A significant increase in yield strength was shown vs unexposed material. At 1200°F test temperature, sufficient ductility returned to the samples previously exposed to "C-sodium" to allow measurement of the tensile properties. The values showed higher tensile and yield strength and lower

ductility vs the air-exposed specimens. Carbon penetrated through the entire cross section of the 1/16-in. sample after 4000 hr, but high carbon content was confined to the outer one-third of the surface. Even at 1200°F, ductility was so low as to render the material useless. Failure occurred due to brittle cracking of the outer carburized case followed by ductile failure of the inner core.

F. Effect of Sodium with Oxygen Impurity on 2 $\frac{1}{4}$ Cr-1 Mo (MSAR)⁸⁰

A 4000-hr exposure in oxygen-contaminated sodium at 1100°F had a greater effect on the room-temperature tensile properties of 2 $\frac{1}{4}$ Cr-1 Mo than on 316 SS or 304 SS. The tensile and yield strengths of 2 $\frac{1}{4}$ Cr-1 Mo were reduced, but the ductility was hardly affected when compared with controls exposed at 1100°F in helium (Table 3.4). Short-time tensile and yield strengths at 1100°F, after aging in oxygen-contaminated sodium, were less than 10% lower compared to 1100°F helium-aged controls, with a correspondingly small increase in ductility. High-oxygen sodium exposure resulted in somewhat greater changes in tensile properties due to decarburization than low-oxygen sodium.

The tensile properties of 2 $\frac{1}{4}$ Cr-1 Mo before and after exposure, reduced to their 25% and 62 $\frac{1}{2}$ % values for tensile and yield strength, respectively, are compared against code allowable tensile and yield stresses in Figs. 3.45 and 3.46. As can be seen in these figures, even after decarburization by high-oxygen sodium, 2 $\frac{1}{4}$ Cr-1 Mo strength values do not fall below code allowable stresses.

British workers⁹⁶ have developed tensile property data for as-received and decarburized 2 $\frac{1}{4}$ Cr-1 Mo at various temperatures and levels of uniform carbon content. These results are presented in Table 3.5. They show the same trends, established by the MSAR data, i.e., decreases in strength with corresponding increases in ductility when original 0.12% C specimens were decarburized to 0.03% C. However, further decarburization to 0.015% showed a reversal in the above trends with strength increasing slightly but with little significant effect on ductility. The latter result is at variance with the MSAR results and is probably due to the interaction effects of spheroidization in decreasing strength in higher carbon specimens after aging. This effect decreases with decreasing carbon content, nearly disappearing at low levels such as 0.015% C.

G. Effect of Nonisothermal Sodium System on 316 SS and a Typical Low Alloy Ferritic Steel^{74, 106}

Table 3.6 shows tensile property data taken on heat exchanger tubes before and after 10,000 hr of exposure in sodium where some mass transfer is occurring

between ferritic steel and austenitic stainless steel at different temperatures. The tensile property results show the effects of carburization of the 316 SS and decarburization of the low-alloy steel due to ΔT , dissimilar metal, and oxygen in the sodium. Although the low-alloy steel was $1\frac{1}{4}$ Cr- $1\frac{1}{2}$ Mo, similar effects could be expected with $2\frac{1}{4}$ Cr-1 Mo. Type 316 SS in the hot leg generally increased in strength and decreased in ductility whereas $1\frac{1}{4}$ Cr- $1\frac{1}{2}$ Mo in the cold leg decreased in strength and increased in ductility. Changes in properties varied directly with the metal operating temperature. Tensile data^{57, 108, 145, 150a, 165} for $2\frac{1}{4}$ Cr-1 Mo suggest a linear decrease of both ultimate and yield strength with decreasing carbon content.

3.3.2 Effects of Sodium on Creep and Rupture Strength

A. 316 SS MSAR Data⁸⁰

Some 1/16-in. thick samples of 316 SS were tested at 1200°F in environments described in Table 3.7 for creep strength after 1200°F pre-exposure for 4000 hr. Pre-exposure environments were high-oxygen sodium, helium, and high-carbon sodium. Controls which were not pre-exposed to a high-temperature environment were also tested. Figs. 3.47, 3.48, and 3.49 present test results for the pre-exposure and test environment combinations described in Table 3.7. Fig. 3.49 and Table 3.7 present extrapolated values for 1% creep in 100,000 hr based on data in Figs. 3.47 and 3.48. Fig. 3.49 also shows, for comparison purposes, ASTM average creep rate data taken in air. Creep strength data obtained by MSAR under the various described conditions all were higher than the ASTM average, with the 316 SS specimens with the longest exposure to high-carbon sodium exhibiting the highest creep strength. Most of the above data represented average values from specimens tested in triplicate.

Data on 1200°F creep-rupture strength were obtained by MSAR⁸⁰ on the same size and heat of 316 SS in the same environments as above. Part of these data are shown in Fig. 3.50 with the curves extrapolated to 100,000 hr. The 100,000-hr extrapolated values are plotted in Fig. 3.51, along with a curve showing ASTM average values vs temperature. All of the noncarbon exposure data are averaged into one value in Fig. 3.51 as the range in actual values was very narrow. No significant difference in rupture strength was found between air and high-oxygen samples. Specimens exposed to high carbon showed higher strengths than specimens exposed in other media including the ASTM air exposure. MSA noncarbon-containing sodium media values, including air and helium, were slightly below ASTM values.

So-called 1200°F “design” curves were developed from the rupture strength test data taken in high-oxygen sodium and high-carbon sodium without and with 4000-hr, 1200°F pre-exposure. These curves were based on results obtained from “numerous” uniaxially loaded tests. Due to doubts that might be raised on the sufficiency of data, on sample thickness and heat variation, and particularly on precise characterization of the sodium environment, it is unlikely that these curves can be designated design curves.

These “design” curves are presented in Figs. 3.52, 3.53, 3.54, and 3.55. Pre-exposure in high oxygen sodium somewhat increases creep strength at moderate rates but has no significant effect on high creep rates or rupture strength. Exposure to high-carbon sodium increases creep and rupture strength but reduces high-temperature ductility. Pre-exposure in high-carbon sodium enhances these effects.⁸⁰ These effects would diminish on thicker specimens.

The later MSAR work^{135, 136} on 316 SS shows behavior similar to 304 SS as a result of prior exposures, with some weakening as compared with as-received material tested in air. In both materials, testing in clean sodium results in lowered properties when compared with tests in air. All exposures except 1400°F oxygen contaminated Na result in transgranular (shear) failure of 316 SS. Exposure at 1400°F to oxygen contaminated sodium results in intergranular failure due to surface oxidation. On the whole, the pre-exposure conditions have a relatively minor effect on stress-rupture behavior.

Exposure of 316 SS for 750 hr to high-carbon sodium at 1200°F results in increased rupture strength and decreased ductility due to carburized case and enhanced carbide precipitation. Failure begins by brittle cracking of the case which proceeds through the region of carbide grain boundary network. In the low-carbon region of the base metal, failure shifts from brittle (intergranular) to ductile (transgranular). In 316 SS there is no evidence of carbon migration from the case into the base metal and no evidence of ferrite formation.

B. 304 SS (MSAR)¹⁰⁰

The 1200°F creep-rupture strength of 304 SS after exposure to oxygen and carbon contaminated sodium and tested in similar environments has been determined by MSA.¹⁰⁰ These data are compared against 304 SS exposed and tested in air. Fig. 3.56 shows a comparison of rupture strength at 1200°F between as-received material tested in air vs high-oxygen (300 ppm) sodium. High-oxygen sodium test results were approximately one-half lower in rupture life. Fig. 3.57 presents rupture strengths on variously pre-exposed (4000 hr) material in juxtaposition with the plots on as-received material from Fig. 3.56. The pre-exposure/test media combinations were helium/air, high-oxygen sodium/air, high-oxygen sodium/high-oxygen sodium. The results from pre-exposure

showed no significant change from results on as-received materials tested in high-oxygen sodium; that is, they showed a decrease in rupture strength vs the air-tested control.

Creep-strength results were presented in the same manner as rupture strengths, above, except no as-received creep data for high-oxygen sodium were available. Fig. 3.58 shows comparative creep rate vs stress plots of 304 SS tested in air, and 304 SS pre-exposed and tested in various environmental combinations. Creep rates at the same stress were slightly greater for pre-exposed material, particularly in high-oxygen sodium, than original material tested in air.

In later work at MSAR,^{135,136} 304 SS exposed at 1200°F to high-oxygen sodium, high-carbon sodium, or helium environments shows lower stress-rupture strength than unexposed material when tested in air. When tested in sodium, strengths are even lower. Exposure at 1400°F to the same environments results in lower strength. Failure mode is by shear at high stresses and intergranularly at low stresses for material with no prior exposure tested in clean sodium. Ductility decreases with time. These are manifestations of carbide sensitization. This is counteracted by surface depletion of Ni and Cr. Material exposed to 1200°F He and tested in sodium shows intergranular failure. Prior oxidation appears to counteract strengthening due to carbide sensitization.

Pre-exposure at 1400°F results in properties below those of as-received material which is attributed to carbide re-solution. Pre-exposure at 1400°F to oxygen contaminated sodium results in poorest properties owing to oxidation product (possibly Na_2CrO_4) in grain boundaries.

In 304 SS, exposure to carbon contaminated sodium results in little if any strengthening and a severe reduction of ductility. This results from the combination of a carbide case, a ferrite grain boundary network, and continued carbon diffusion.

C. Austenitic Stainless Steels (AI)

Atomics International (AI) subjected tubular specimens of 304 SS and 316 SS to biaxially loaded stress-rupture tests in low (4 to 11 ppm) oxygen sodium at 1300 and 1400°F temperatures.^{107,110,111} Pre-selected stress levels used in the tests were made on the basis of normal rupture life values of 100 to 1000 hr. Rupture life values obtained in these tests are shown in Table 3.9 and Fig. 3.59. Rupture life, time to rupture at a given stress, was lowered due to sodium exposure, compared to helium, e.g., one-third of the life for sodium vs helium at 6800 psi and 1400°F. Biaxial rupture life values reported are ap-

proximately equal to uniaxial values.¹¹⁰ Subsequent analyses of the sodium-exposed specimens after testing showed a 50% lowering in carbon content (0.06 to 0.03%). This is attributed to the carbon-gettering action of zirconium, which was also in contact with the sodium in the test capsule. This decarburization might partially explain the lowering of rupture life by nearly "pure" sodium (<11 ppm O₂) in two ways: (1) the lower carbon 304 SS will have somewhat less rupture life; (2) lower carbon promotes sigma formation, which leads to lower rupture life (extensive sigma was found in the matrix of the sodium exposed samples). Sigma also could be expected in the helium-exposed samples but no information on this was reported. The temperatures are also higher than the 1200°F anticipated in the LMFBR primary.

In later work in which the Zr gettering was eliminated, AI studies¹¹⁰ of the biaxial stress-rupture behavior of one heat of type 304 stainless steel and one heat of type 316 stainless steel in high-purity sodium (O₂ ~10 ppm) and helium (99.99% pure) revealed the following.

1. No significant degradation of mechanical properties was found in a high-purity sodium environment for testing times up to 3000 hr at 1200°F and 1000 hr at 1400°F.
2. The internal stresses due to 10 to 15% cold work did not induce intercrystalline penetration by liquid sodium.
3. Use of 10 to 15% cold work reduced long-term stress-rupture properties to below the properties for annealed material for both type 304 and type 316 stainless steel at 1200 and 1400°F. The 10 to 15% cold work also decreased the activation energy for creep; the effect was more pronounced for type 304 stainless steel.
4. Cold work enhanced precipitation within grains and weakened the grain boundary, promoting extensive grain-boundary void formation, due to grain-boundary sliding.
5. The formation of sigma phase in tested specimens was attributable to the presence of cold work and deformation during the stress-rupture test at high temperature.
6. Above 1200°F, 10 to 15% cold-worked type 316 stainless steel exhibited significantly higher long-term rupture strength than cold-worked type 304 stainless steel, and comparable ductility (diametral strain). At 1000°F, cold-worked type 304 stainless steel had a higher rupture strength.

These results are summarized in Table 3.8 and Figs. 3.59 and 3.60.

In other AI work,¹¹¹ stress-rupture tests were run on internally heated and pressurized thin-walled tubes of 304 SS immersed in high-purity Na flowing at 20 ft/sec with heat flux of 10^6 Btu/ft²-hr through the wall. Heat flux was applied in three modes: steady; 6-minutes on/6-minutes off; and 1-minute on/1-minute off. NaK or He and flowing Na have no effect on creep behavior. Cyclic heating produced large changes in microstructure, strain profile, total strain, and stress-rupture life. Results were explained on the basis of a ratcheting mechanism involving primary creep.

Böhm and Schneider¹⁸⁴ studied the effect of sodium on the creep and stress rupture properties of two austenitic steels at 700°C. Tubes of the two alloys were filled with sodium and tested under a uniaxial stress. Although the secondary creep rate was not affected by the sodium, the tertiary creep stage was shortened, resulting in a reduction of fracture strain. Figs. 3.61 through 3.66 summarize their results.

D. $2\frac{1}{4}$ Cr-1 Mo - MSA Data^{79, 80}

The minimum creep rates of $2\frac{1}{4}$ Cr-1 Mo steel for tests in air, helium, and sodium with 30 and 300 ppm oxygen impurity are shown in Fig. 3.67. Exposure to sodium increased creep rates due to decarburization. Pre-exposure to sodium had the most effect in increasing creep. The oxygen level in the sodium had no apparent effect on the creep rates in sodium. Table 3.10 presents extrapolated values for creep strength for 1% creep in 100,000 hr at 1100°F based on data in Fig. 3.67. These data are especially tentative since they are based on extrapolations which are beyond those usually acceptable.

Creep-rupture strength values at 1100°F were obtained for the same environments as creep rates. As Fig. 3.68 shows, the rupture strength is somewhat higher for material tested in air or helium than in sodium, a result of decarburization in sodium. Fig. 3.69 and Table 3.10 show creep and creep-rupture data, respectively, for $2\frac{1}{4}$ Cr-1 Mo pre-exposed in low and high oxygen content sodium at 1100°F for 4000 hr vs helium pre-exposure as a control. The lowest creep-rupture strength was exhibited by specimens pre-exposed to sodium, with the lowest for sodium with high oxygen (Fig. 3.69). The creep-rupture data in Fig. 3.69 were extrapolated to rupture stress after 100,000 hr and these data were plotted in Fig. 3.70. Fig. 3.70 also shows average Timken extrapolated data⁵³ with an indication of its range along with "design" data based on 60% of ASTM average values. This figure shows that exposure to sodium with oxygen produced 100,000-hr rupture stresses below air values (Timken) and in some cases (high-oxygen pre- and test-exposure) below allowable "design" stress.

Figs. 3.71 and 3.72 show, in a different form (stress vs time at various strain levels including rupture), creep rates and rupture strengths for all high-oxygen sodium exposures of $2\frac{1}{4}$ Cr-1 Mo described. Comparative creep rates and rupture stresses for nonsodium exposures were not presented. Figs. 3.71 and 3.72 are called "design curves," but, due to limitations in number of specimens tested and heats (one), caution must be exercised in applying these curves for design purposes. Further, the sample thicknesses were 1/16 in., which, after 10,000 hr pre-exposure in sodium, resulted in nearly uniform levels of decarburization, a situation which will not exist on thicker walls such as LMFBR piping.

E. $2\frac{1}{4}$ Cr-1 Mo - British Data^{96, 108}

British data on tensile properties of decarburized $2\frac{1}{4}$ Cr-1 Mo have been previously presented in Table 3.5. This same source has conducted short-term rupture tests on 0.015% C specimens which showed a drop in rupture strength of 20% in 550 hr at 1000°F vs the undecarburized specimen (0.12% C). Time to rupture at a 27,000 psi stress level at 1000°F was considerably lower for the 0.015% C specimen (500 vs 1700 hr).

F. Duplex Impurity Exposures - MSAR Data^{135, 136}

Exposure to oxygen contaminated sodium followed by exposure to carbon contaminated sodium or vice versa at 1200°F did not result in synergistic effects. However, prior oxidation slowed down subsequent carburization while prior carburization had the same effect on oxidation. Tensile properties for specimens with dual exposures are basically similar when account is taken of differences in carbon pickup. Stress rupture behavior of specimens with exposure to O-C vs C-O impurities is consistent with the type of exposure. O-C results in less carburization, more ferrite, and a thinner case; the result is a weaker material.

3.3.3 Effects of Sodium on Impact Strength (MSAR)

A. 316 SS (MSAR)⁸⁰

Decreases in room-temperature toughness of 316 SS occur during prolonged heating, as pointed out in Section 2.5.2 which discusses impact properties after elevated-temperature aging. Elevated-temperature aging effects are compounded by the effects of exposure to oxygen and carbon contaminated sodium. Tests to evaluate specimens of 316 SS in triplicate in 1200°F environments were conducted by MSAR. Fig. 3.73 shows the results on 316 SS room-temperature impact values of various exposure times in helium (closely equivalent to the air

aging environment in Section 2.5.2), high-oxygen (300 ppm) sodium, and high-carbon (37 ppm) sodium at 1200°F. The effects of 1200°F aging on decreasing the room-temperature impact values of 316 SS are most pronounced in the first 500 hr and then decrease to a low rate of change with time. After about 4000 hr, 316 SS has lost 65 and 82% of its impact strength in oxygen-and carbon-sodium, respectively; but of these losses, 45% is accountable to metallurgical aging effects as evidenced by the helium exposure result (45% loss). Metallurgical examination of the high-carbon specimens revealed brittle fracture failure due to carburization. It is thought⁸⁰ that if the specimens were stressed during high-carbon sodium exposure, the carburization effects would have been even greater. Obviously, impact losses due to high-oxygen and carbon contamination of sodium are potentially so severe that oxygen and carbon contamination must be carefully controlled, even in systems with heavy-wall steel pipe. The LMFBR is designed for >250,000 hr of life.

B. 304 SS (MSAR)¹⁰⁰

MSAR studied the effects of exposure of 304 SS (0.098-in. thick) to high-oxygen sodium and high-carbon sodium at 1200°F on room-temperature toughness as measured by the Charpy Keyhole impact test. These results, plotted in Figs. 3.74 and 3.75, can be compared against impact values taken after exposure to 1200°F helium shown in the same figures. The following conclusions are made from these data.

1. Impact tests run concurrently with sodium exposure tests showed an approximately 20% loss in impact strength for 304 SS exposed to helium for 4000 hr at 1200°F, compared to zero exposure time. This loss in room-temperature toughness is attributable to the effects of 1200°F aging (see Section 2.5.2).
2. An approximately 45% loss in toughness occurred in 304 SS exposed to high-oxygen sodium for the same time, at least one-half of which loss can be attributed to 1200°F aging increasing the sigma content as indicated by the helium results, above. Most of the loss in toughness in high-oxygen sodium occurred as a result of the first 800 hr of exposure, whereas the losses occurring during 4000 hr of helium aging occurred at nearly constant rate throughout.
3. Eighty-six percent of the initial toughness was lost as a result of the 4000-hr (1200°F) exposure in high-carbon (30 ppm) sodium, an effect due to carburization and sigma formation.

Sixty-four percent of the 86% total loss occurred during the first 500 hr of exposure. Again, as with 316 SS, it is clear from these data that impact losses on 304 SS are potentially very severe due to oxygen, and particularly, carbon contamination of sodium. Oxygen and carbon levels must be minimized by careful control, even in systems with heavy-wall austenitic steels.

C. $2\frac{1}{4}$ Cr-1 Mo (MSAR)^{79, 80}

Elevated temperature aging in air, described in Section 2.5.2, was not detrimental to the toughness of $2\frac{1}{4}$ Cr-1 Mo. Toughness was evaluated by Charpy Keyhole notch impact tests. Similarly, exposure to high-oxygen sodium produced no deterioration in toughness, but rather a slight increase, as shown in Fig. 3.76, MSAR data on environmental effects on impact properties after extended exposure times at 1200°F. This increase in toughness was probably due to both decarburization and spheroidization. Exposure to high-carbon sodium resulted in a lowering in toughness after an initial rise which probably was caused by the slow carburization after initial spheroidization.

3.3.4 Effects of Sodium on Fatigue Strength

A. 316 SS

Fatigue studies on 316 SS during exposure to sodium have been conducted by MSAR⁸⁰ and Thiruvengadem.¹⁰⁹ MSAR⁸⁰ strain-controlled fatigue tests were performed in at least triplicate at 180 cycles/hr at 1200°F on 1/16-in. specimens at 2, 1, and 1/2% maximum strain levels. Tests were run on pre-exposed 316 SS (low- and high-oxygen sodium, high-carbon sodium) as well as on as-received material. The test environments were: air, helium, low- (30 ppm) and high-oxygen (300 ppm) sodium, and high-carbon (30 to 50 ppm) sodium. The combinations of pre-exposure and test variables and the resulting data are described in Table 3.11.

Fig. 3.77 presents graphically the as-received (no pre-exposure) testing data in Table 3.11. These data show that exposure to high- and low-oxygen sodium during test increases fatigue life vs air. Tests in high-carbon sodium and air produced about the same fatigue behavior. At low strains (<0.5%), high-oxygen sodium results showed a decrease in fatigue life vs air. The "corrected for carbon" annotation in Fig. 3.77 is a result of finding random carburization in the low- and high-oxygen sodium environments. MSAR had data which enabled them to correct the fatigue results of these tests. Fig. 3.78 used data from Table 3.11 calculated in the form of plastic strain plotted vs cycles to

failure. The results were extrapolated back to the calculated "1/4 cycle fracture ductility point." In this concept the tensile test is regarded as a quarter cycle of a fatigue test and the line of strain can be expected to intersect the life $N=1/4$ at a value of strain which is equal to the strain at fracture.

Thiruvengadem¹⁰⁹ employed magnetic striction apparatus to produce high-cycle fatigue data on vibrating rod-type 316 SS specimens immersed in sodium. Tests were performed in oxygen-sodium (30 to 100 ppm oxygen levels) at 1000 and 1500°F at ~14,000 cycles/sec. The results of this work are shown in Figs. 3.79 and 3.80. Unfortunately, no air control was tested and oxygen-sodium exposure times were too short to enable any environmental effects on fatigue life to be adequately evaluated. As expected, fatigue properties at 1500°F were considerably lower than at 1000°F.

B. 304 SS (MSAR)¹⁰⁰

Strain-controlled fatigue studies, of the same type as MSAR conducted on 316 SS and described above, were conducted on 304 SS. These long-cycle tests were performed in at least triplicate at 180 cycles/hr at 1200°F on 1/16-in. specimens at 2, 1, and 1/4% maximum strains. Tests were run on as-received and pre-exposed 316 SS (low- and high-oxygen sodium, high-carbon sodium). The test environments were: air, helium, high-oxygen (300 ppm) sodium, and high-carbon (30 to 50 ppm) sodium. The combination of pre-exposure and test variables and the resulting data are presented in Tables 3.12 and 3.13 and summarized as strain vs cycles to failure in Figs. 3.81 and 3.82. These data show that tests in high-oxygen sodium (corrected for carbon) reduce fatigue strength compared to tests in helium. The air test vs oxygen-sodium showed lower fatigue strength in air above 0.8% strain but less below this strain; helium strength was higher than both. However, fatigue properties resulting from testing in 1200°F high-carbon sodium environment when compared to testing in air (Fig. 3.82) showed little significant difference in fatigue life. The test in helium provided significantly higher fatigue life than either air or high-carbon sodium. Apparently, the oxidation effects of air and the carburization effects of high-carbon sodium reduced fatigue life vs the inert helium test environment.

C. 2¹/₄ Cr-1 Mo (MSAR)

Strain-controlled, long cycle (180 cycles/hr) fatigue strength studies at 1100°F on 2¹/₄ Cr-1 Mo were conducted by MSAR^{79, 80} in four different environments: air, helium, and low- and high-oxygen sodium. The work was conducted, as for 316 SS above, at strain levels of 2.2%, 1.0%, and 0.56%. The test results are plotted in Fig. 3.83. The longest fatigue life was produced in high-oxygen

sodium followed very closely by helium and low-oxygen sodium. In summary, three of these environments gave almost the same fatigue properties. The fatigue life in the air control environment was significantly lower than in the above three. This large difference was undoubtedly accentuated by the relatively high oxidation rates of the $2\frac{1}{4}$ Cr-1 Mo in 1100°F air.

TABLE 3.1 — EFFECTS OF OXYGEN IN FLOWING SODIUM ON CORROSION RATE
OF STAINLESS STEEL^{50, 68, 74}

Material	Temperature, °F	Surface Removal Rate, mils/yr				
		At below 5 ppm Oxygen	At 10-15 ppm Oxygen	At 25-30 ppm Oxygen	At 45-50 ppm Oxygen	At 100-300 ppm Oxygen
316 SS ⁶⁵	1100	—	1/4	—	1	—
	1200	—	1/2	—	2	—
	1300	—	1	—	4	—
304 SS ⁶⁷	1000-1020	—	0.05	—	0.25	—
316 SS ⁶⁸	1200	—	0.4	—	1.9	—
316 SS ²⁹	1200	0.15-0.2	0.5-0.7	1.5-2	—	—
Stainless Steel ⁷⁰ (16 Cr, 15 Ni, 3 Mo, Nb)	1200	—	—	—	0.3	1.8
18 Cr-Ni Steel ^{50, 71}	1022	—	—	—	0.24	2.4

TABLE 3.3 — TENSILE PROPERTIES FOR TYPE 316 STAINLESS STEEL⁸⁰

Exposure Environment 1200°F (1033°F)	Test Environment Test Temperature	Tensile Strength (psi)	Average Elongation (%)	0.2% Offset Yield Strength (psi)	Average Yield (psi)	Elongation (%)	Average Elongation (%)	Reduction of Area (%)	Average Reduction of Area (%)
None	Helium RT	89,500	83,000	44,700	40,550	62	54	54	0
None	Helium RT	88,000	83,000	44,400	40,550	60	54	54	0
Low-oxygen Na	Helium RT	83,200	82,500	44,000	43,300	47	48	48	47
Low-oxygen Na	Helium RT	82,000	82,500	43,700	43,300	48	48	48	47
Low-oxygen Na	Helium RT	82,500	82,500	42,000	43,300	46	48	48	47
Low-oxygen Na	Helium RT	81,400	82,500	43,700	43,300	47	48	48	47
Low-oxygen Na	Helium RT	82,500	82,500	43,000	43,300	49	48	48	47
Low-oxygen Na	Helium RT	82,500	82,500	43,000	43,300	49	48	48	47
High-oxygen Na	Air RT	83,200	83,000	40,300	34	34	32	32	32
High-oxygen Na	Air RT	83,900	83,000	41,200	34	34	34	34	32
High-oxygen Na	Air RT	84,000	83,000	37,000	34	34	34	34	32
High-oxygen Na	Air RT	85,000	83,000	37,000	34	34	34	34	32
High-oxygen Na	Air RT	85,600	83,000	41,250	34	34	34	34	32
High-oxygen Na	Air RT	82,000	83,000	41,700	34	34	34	34	32
Helium†	Helium RT	87,000	87,300	38,500	36,100	55	54	54	50
Helium†	Helium RT	87,700	87,300	27,300	36,100	53	54	54	50
Helium†	Helium RT	87,200	87,300	40,000	36,100	53	54	54	50
High-carbon Na	Helium RT	67,800	69,767	23,750	27,833	<15	<1	<1	5.5
High-carbon Na	Helium RT	66,500	69,767	27,500	27,833	15	1	1	7.4
High-carbon Na	Helium RT	78,100	69,767	27,250	27,833	2.45	1	1	8.5
High-carbon Na	Helium RT	68,800	69,767	57,750	27,833	2.05	1	1	9.2
High-carbon Na	Helium RT	68,400	69,767	62,200	27,833	1.25	1	1	7.0
High-carbon Na	Helium RT	69,000	69,767	58,500	27,833	2.15	1	1	7.0
Helium†	Air RT	87,100	87,000	40,250	40,783	53	54	54	51
Helium†	Air RT	86,400	87,000	38,000	40,783	50	54	54	51
Helium†	Air RT	87,500	87,000	41,200	40,783	49	54	54	51
None	Air 1200°F	49,400	50,280	26,100	26,020	43	43	43	40
None	Air 1200°F	50,250	50,280	25,000	26,020	41	43	43	40
None	Air 1200°F	51,200	50,280	26,300	26,020	47	43	43	40
None	Helium-1200°F	47,200	47,770	26,000	25,430	46	47	47	55
None	Helium-1200°F	47,500	47,770	26,300	25,430	52	47	47	55
None	Helium-1200°F	48,600	47,770	25,300	25,430	44	47	47	55
Low-oxygen Na	Helium-1200°F	48,850	48,580	24,700	25,170	39	44	44	48
Low-oxygen Na	Helium-1200°F	46,900	48,580	24,800	25,170	42	44	44	48
Low-oxygen Na	Helium-1200°F	50,000	48,580	26,200	25,170	41	44	44	48
Low-oxygen Na	Helium-1200°F	49,100	48,470	26,700	25,870	41	44	44	54
Low-oxygen Na	Helium-1200°F	46,400	48,470	25,200	25,870	46	44	44	54
Low-oxygen Na	Helium-1200°F	49,900	48,470	25,650	25,870	46	44	44	54
High-oxygen Na	Helium-1200°F	46,700	47,867	24,400	24,300	27	28	28	40
High-oxygen Na	Helium-1200°F	47,800	47,867	24,000	24,300	31	28	28	40
High-oxygen Na	Helium-1200°F	49,100	47,867	24,200	24,300	25	28	28	40
High-oxygen Na	Helium-1200°F	49,500	49,083	23,300	24,300	25	23	23	33
High-oxygen Na	Helium-1200°F	49,250	49,083	25,000	24,300	24	23	23	33
High-oxygen Na	Helium-1200°F	48,500	49,083	24,000	24,300	21	23	23	33
Helium†	Helium-1200°F	48,000	55,750	23,700	26,293	27	47	47	48
Helium†	Helium-1200°F	54,250	55,750	27,600	26,293	24	47	47	48
Helium†	Helium-1200°F	65,000	55,750	27,750	26,293	26	47	47	48
High-carbon Na	Helium-1200°F	57,300	50,200	39,700	40,167	2.15	2.2	2.1	3.0
High-carbon Na	Helium-1200°F	54,300	50,200	40,500	40,167	1.75	2.2	1.7	3.0
High-carbon Na	Helium-1200°F	57,000	50,200	40,300	40,167	2.75	2.2	2.7	3.0
High-carbon Na	Helium-1200°F	59,000	57,875	38,300	40,417	2.35	2.0	1.5	2.0
High-carbon Na	Helium-1200°F	56,375	57,875	41,250	40,417	1.75	2.0	1.7	2.0
High-carbon Na	Helium-1200°F	58,250	57,875	42,000	40,417	2.05	2.0	2.0	2.0
Helium†	Helium-1200°F	48,100	46,867	23,700	25,000	44	23	22	64
Helium†	Helium-1200°F	46,400	46,867	24,400	25,000	31	23	27	64
Helium†	Helium-1200°F	46,100	46,867	23,500	25,000	26	23	22	64

RT - Room Temperature

* Exposed and then cooled with the Helium-1200°F exposure

† Exposed and then cooled with the high-carbon Na exposure

§ Failed at grain boundary

** Not washed prior to test in Helium exposure; failed with alcohol prior to test

TABLE 3.4 — TENSILE PROPERTIES FOR 2¹/₄ Cr-1 Mo STEEL⁸⁰

Exposure Environment 1100°F - 4000 hr	Test Environment Test Temperature	Tensile Strength (psi)	Average Tensile Strength (psi)	0.2% Offset Yield Strength (psi)	Avg Yield Strength (psi)	Elongation (%)	Average Elongation (%)	Reduction of Area (%)	Average Reduction (%)
None	Air - RT	68,050	70,710	42,500	37,060	30	20	-	-
↕	↕	68,350		43,000		31		-	
		72,900		32,450		29		-	
		73,400		34,150		29		-	
None	Air - RT	70,850		33,200		28		-	
High-oxygen Na	Helium - RT	61,600	61,437	28,600	26,950	23	28	62	60
↕	↕	69,280		25,800		33		65	
		57,910		24,400		28		60	
		59,550		28,300		21		55	
		59,760		27,300		28		65	
High-oxygen Na	Helium - RT	60,520	66,000	27,300	28,700	32	26	53	61
↕	↕	67,300		29,700		28		57	
Helium*		64,500		26,900		23		65	
Helium*	Helium - RT	66,200		29,700		28		62	
None	Air - 1100	49,170	55,123	25,800	27,090	28	31	64	61
None	Air - 1100	56,800		28,800		32		57	
None	Air - 1100	59,400		29,370		32		62	
None	Helium - 1100	39,700	38,547	25,000	23,700	26	33	63	60
None	Helium - 1100	38,120		22,600		31		44	
None	Helium - 1100	37,820		23,500		41		74	
Low-oxygen Na	Helium - 1100	26,620	26,113	17,200	17,100	45	45	54	50
↕	↕	25,250		-		46		55	
		26,580		16,800		47		41	
		26,160		17,300		41		50	
		25,800		16,100		40		65	
Low-oxygen Na	Helium - 1100	26,410	27,137	17,900	17,833	42	40	56	59
↕	↕	23,200		19,400		38		57	
High-oxygen Na	Helium - 1100	23,800	21,720	18,300	16,000	46	41	64	61
High-oxygen Na	↕	20,620		14,700		36		63	
High-oxygen Na		20,740		15,000		39		57	
High-oxygen Na†		23,400	22,567	14,500	13,850	37	37	62	65
High-oxygen Na†		22,700		-		36		67	
High-oxygen Na†		22,000		13,200		37		66	
Helium*		20,800	25,730	13,500	16,707	33	37	81	77
Helium*		28,200		18,000		30		73	
Helium*	Helium - 1100	28,200		18,900		47		76	

RT = Room Temperature

*Exposure medium was high oxygen Na₂S₂O₈ solution

†Not welded prior to testing; other specimens were welded (thoroughly and after prior testing)

TABLE 3.5 — MECHANICAL PROPERTIES OF DECARBURIZED
2¹/₄ Cr-1 Mo STEEL⁹⁶

(Material decarburized in wet H₂ and heat treated before test.
Heat Treatment: Normalize at 1742°F for 1/2 hr; temper at
1292°F for 1/2 hr.)

Test Temp. °F	Condition	Y.S. ksi	U.T.S. ksi	Elong. %	R.A. %
R.T.	As rec'd 0.12% C	52.0	80.0	31	72
	Decarbd.to 0.03%C	34.2	51.4	43	80
	Decarbd.to 0.015%C	37.6	53.3	46	84
572	0.12%C	47.2	72.2	26	49
	0.03%C	23.4	49.6	35	78
	0.015%C	31.2	57.6	30	70
752	0.12%C	52.8	75.6	24	52
	0.03%C	22.8	54.6	33	80
	0.015%C	22.4	56.0	30	72
842	0.12%C	52.0	75.4	24	45
	0.03%C	22.4	52.4	33	76
	0.015%C	27.2	57.2	27	76
1000	0.12%C	46.4	61.6	26	65
	0.03%C	22.8	46.0	33	80
	0.015%C	20.8	45.8	33	80

TABLE 3.6 — MECHANICAL PROPERTIES OF
HEAT EXCHANGER TUBES BEFORE AND AFTER
10,000 HR OF OPERATION¹⁰⁶

	Average metal operating temp., °F	Elongation		Outside surface hardness, DPH (5-kg load)	Tensile* strength, 1000 psi	
		Per- cent	Gauge length, in.		Yield	Ultimate
Intermediate Heat Exchanger (Type 316 Stainless Steel)						
As received†		52	3.5	189	48.0	90.2
After use†	1000	44	3.5	207	49.6	95.1
After use†	760	45	3.5	201	47.6	94.5
Superheater (Type 316 Stainless Steel)						
As received‡		49	5.2	176	51.4	90.1
After use§	960	42	5.2	195	51.9	92.4
After use§	870	41	5.2	208	50.1	88.2
Tempering Heat Exchanger (1¼% Cr-½% Mo)						
As received‡		26	3.7	184	57.4	73.6
After use¶	840	35	3.7	130	35.6	66.0
After use¶	580	34	3.7	128	39.2	66.4
Evaporator (1¼% Cr-½% Mo)						
As received‡		22	5.1	167	47.9	72.5
After use¶	680	30	5.1	138	37.2	62.2
After use¶	540	31	5.1	130	40.3	61.6
After use¶	500	30	5.1	131	42.2	64.4

*Strain rate: 0.020 in./min to yield,
0.10 in./min to ultimate.

†Average of three samples.

‡Average of six samples.
§One sample.

¶Average of two samples.

TABLE 3.7 — CREEP DESIGN DATA FOR TYPE 316
STAINLESS STEEL IN VARIOUS ENVIRONMENTS⁸⁰

Exposure Environment 4000 hr at 1200°F	Test Environment at 1200°F	Stress to Produce Creep Rate of 1% in 100,000 hr (psi)
None	Helium	8,200
None	Air	10,500
None	Low-oxygen sodium	8,500
None	High-oxygen sodium	8,500
None	High-carbon sodium	10,000
High-oxygen sodium	High-oxygen sodium	8,500
High-oxygen sodium	Air	8,500
High-oxygen sodium	Helium	10,400
Helium	Air	8,600
High-carbon sodium	Air	11,800
High-carbon sodium	Helium	13,200
High-carbon sodium	High-carbon sodium	17,000

TABLE 3.8 — BIAXIAL STRESS-RUPTURE DATA FOR TYPE 304
STAINLESS STEEL¹¹⁰

(Heat No. 20013)

Alloy Condition	Environment	Temperature (°F)	Specimen Number	Hoop Stress (psi)	Rupture Time (hr)	Diastical Strain (maximum) (%)
10 to 15% Cold Work (As-received)	Sodium (O ₂ ~ 10 ppm)	1100	4-64	11,000	201	5.6
			-63	11,000	201	4.9
			-62	11,000	222	4.7
			-61	9,500	320	5.2
			-60	9,500	332	4.9
			-59	9,500	296	3.8
			-58	8,000	512	5.0
			-57	8,000	512	5.5
			-56	8,000	155	3.5
			-55	6,800	513	6.1
			-54	6,800	632	3.2
			-53	6,800	775	5.3
			1-100	6,000	425	5.7
			-99	6,000	596	3.8
			-98	6,000	849	5.4
			-97	4,700	1102	5.4
			76	4,700	1119	6.0
			-95	4,700	1488	5.9
			-94	4,100	1785	4.2
			93	4,100	1990	6.7
			-92	4,100	2192	6.4
			91	3,500	3032	7.3
			-90	3,500	2864	5.7
			-89	3,500	2915	5.8
Annealed	Sodium (O ₂ ~ 10 ppm)	1400	A4-139	11,000	230	13.8
			-138	11,000	230	10.5
			-137	11,000	250	12.5
			-136	9,500	552	11.1
			-135	9,500	468	11.3
			-134	9,500	502	12.0
			-133	8,000	1260	8.7
			-132	8,000	1306	7.4
			-131	8,000	1094	8.2
			-130	7,200	2170	5.8
			129	7,200	1834	5.8
			-128	7,200	2073	6.7
10 to 15% Cold Work (As-received)	Sodium (O ₂ ~ 10 ppm)	1200	4-112	35,000	273	Burst
			-111	35,000	297	Burst
			-110	35,000	297	Burst
			-109	30,000	765	3.0
			-108	30,000	743	Burst
			-107	30,000	729	2.8
			-106	23,000	1305	2.7
			-105	23,000	2030	2.6
			-104	23,000	2030	2.7
			-103	19,000	3400	3.3
			-102	19,000	3273	3.3
			-101	19,000	3360	3.0

TABLE 3.8 — (CONTINUED)
 BIAxIAL STRESS-RUPTURE DATA FOR TYPE 304 STAINLESS STEEL
 (Heat No. 20013)

Alloy Condition	Environment	Temperature (°F)	Specimen Number	Hoop Stress (psi)	Rupture Time (hr)	Diametral Strain (maximum) (%)
Annealed	Sodium (O ₂ ~ 10 ppm)	1200	A4-24	35,000	58	In Progress
			-23	35,000	58	
			-22	35,000	37	
			-21	30,000	228	
			-20	30,000	228	
			-19	30,000	228	
			-17	25,000	850	
			-16	25,000	875	
			-15	19,000	In Progress	
			-14	19,000	In Progress	
			-13	19,000	In Progress	
			4-76	43,000	737	1.9
			-75	43,000	675	Burst
10 to 15% Cold Work	Sodium (O ₂ ~ 10 ppm)	1100	-74	43,000	494	Burst
			-73	40,000	1328	Burst
			-72	40,000	1174	Burst
			-71	40,000	1125	Burst
			-70	38,000	1477	1.5
			-69	38,000	1560	Burst
			-68	38,000	1760	1.5
			-67	35,000	3300	1.9
			-66	35,000	3373	1.7
			-65	35,000	3206	1.5
			4-36	9,950	263	6.0
			-35	9,950	267	5.1
			-34	9,950	263	5.1
10 to 15% Cold Work	Helium (99.99%)	1400	-33	8,570	344	6.1
			-32	8,570	344	3.4
			-31	8,570	380	4.3
			-30	7,200	531	4.0
			-29	7,200	503	4.7
			-28	7,200	584	5.7
			-27	6,140	789	7.1
			-26	6,140	710	5.1
			-25	6,140	734	5.8
			4-148	35,000	372	In Progress
			-147	35,000	372	In Progress
			-146	35,000	Discontinued	-
10 to 15% Cold Work	Helium (99.99%)	1200	-151	28,000	1257	In Progress
			-150	28,000	1233	In Progress
			-149	28,000	Discontinued	-
			4-145	22,000	2751	In Progress
			-144	22,000	2783	In Progress
			-143	22,000	2937	In Progress
			-142	17,000	In Progress	In Progress
			-141	17,000	In Progress	In Progress
			-140	17,000	In Progress	In Progress
			4-88	75,000	58	Burst
			-87	75,000	45	Burst
			-86	75,000	58	Burst
10 to 15% Cold Work	Sodium (O ₂ ~ 10 ppm)	1000	-85	70,000	105	Burst
			-84	70,000	155	Burst
			-83	70,000	179	Burst
			-82	65,000	285	Burst
			-81	65,000	321	Burst
			-80	65,000	225	Burst
			-79	60,000	550	Burst
			-78	60,000	Discontinued	-
			-77	60,000	680	Burst

TABLE 3.8 — (CONTINUED)
 BIAxIAL STRESS-RUPTURE DATA FOR TYPE 316 STAINLESS STEEL
 (Heat No. 65808)

Alloy Condition	Environment	Temperature (°F)	Specimen Number	Hoop Stress (psi)	Rupture Time (hr)	Diametral Strain (maximum) (%)
10 to 15% Cold Work	Sodium ($O_2 \sim 10$ ppm)	1400	6-12	16,000	152	9.1
			-11	16,000	152	10.8
			-10	16,000	152	11.0
			-9	13,000	405	6.4
			-8	13,000	440	6.9
			-7	13,000	405	5.9
			-6	11,000	689	5.3
			-5	11,000	727	4.9
			-4	11,000	705	5.6
			-3	9,600	968	4.9
			-2	9,600	944	5.4
			-1	9,600	944	4.4
Annealed	Sodium ($O_2 \sim 10$ ppm)	1400	A6-12	16,000	61	26.5
			-11	16,000	61	27.7
			-10	16,000	61	29.4
			-9	13,000	176	28.4
			-8	13,000	152	27.2
			-7	13,000	152	29.9
			-6	11,000	397	22.9
			-5	11,000	397	23.7
			-4	11,000	397	22.0
			-3	9,600	731	19.5
			-2	9,600	681	20.5
			-1	9,600	681	19.1
10 to 15% Cold Work	Helium (99.99%)	1400	6-48	16,000	182	11.6
			-47	16,000	182	11.5
			-46	16,000	182	12.6
			-45	13,000	433	8.4
			-44	13,000	454	7.3
			-43	13,000	469	10.0
			-42	10,300	925	5.3
			-41	10,300	925	4.6
			-40	10,300	963	5.8
			-39	9,300	1123	8.7
			-38	9,300	1142	5.4
			-37	9,300	1549	5.4

TABLE 3.8 — (CONTINUED)
 BIAxIAL STRESS-RUPTURE DATA FOR TYPE 316 STAINLESS STEEL
 (Heat No. 65808)

Alloy Condition	Environment	Temperature (°F)	Specimen Number	Hoop Stress (psi)	Rupture Time (hr)	Diametral Strain (maximum) (%)
10 to 15% Cold Work	Sodium (O ₂ ~ 10 ppm)	1200	6-79	37,000	81	In Progress
			-78	37,000	67	In Progress
			-77	37,000	68	In Progress
			-76	34,000	129	In Progress
			-75	34,000	143	In Progress
			-74	34,000	153	In Progress
			-73	32,000	206	In Progress
			-72	32,000	292	In Progress
			-71	32,000	276	In Progress
			-70	27,000	1822	In Progress
			-50	27,000	1354	In Progress
			-49	27,000	1189	In Progress
Annealed	Sodium (O ₂ ~ 10 ppm)	1200	A6-24	32,000	24	15.9
			-23	32,000	20	17.8
			-22	32,000	19	16.6
			-21	27,000	214	10.4
			-20	27,000	264	10.6
			-19	27,000	259	10.1
			-18	24,000	645	7.4
			-17	24,000	621	7.6
			-16	24,000	681	7.2
			-15	21,000	3054	7.6
10 to 15% Cold Work	Sodium (O ₂ ~ 10 ppm)	1000	-14	21,000	2958	8.5
			-13	21,000	3006	8.5
			6-36	64,000	23	In Progress
			-34	64,000	23	In Progress
			-33	54,000	206	In Progress
			-32	54,000	152	In Progress
			-31	54,000	176	In Progress
			-30	48,000	674	In Progress
			-29	48,000	875	In Progress
			-28	48,000	898	In Progress
			-27	42,000	3970	In Progress
			-26	42,000	4884	In Progress
			-25	42,000	3170	In Progress

TABLE 3.9 — 304 SS CREEP RUPTURE TEST STRENGTH OF ORIGINAL
AND EXPOSED MATERIAL IN HIGH CARBON SODIUM AT 1200 °F¹⁰⁰

Specimen	Exposure Media* 1200 F - 4038 hrs	Stress psi	Rupture Time, hrs	Elong. %	Red. of Area, %	Minimum creep Rate, %/hr
<u>Tested in High Carbon Sodium</u>						
625	None	26,000	49(1)	--	--	-----
701	None	26,000	59	50	39	0.39
624	None	24,000	195	18	21	0.0206
623	None	22,000	390	22	16	0.0150
703	None	19,500	1037	16	13	0.0110
702	None	18,500	3661	13	9	0.0025
615	None	17,500	3228	18	10	0.0027
696	High Carbon Na	22,000	776	12	7	0.0075
698	High Carbon Na	21,000	843	13	9	0.0067
697	High Carbon Na	20,000	1696	12	9	0.0043
<u>Tested in Air</u>						
689	Helium	24,000	117	72	63	0.23
690	Helium	21,000	455	48	64	0.049
691	Helium	19,500	623	55	60	0.040
692	High Carbon Na	24,000	153	8.0	17	0.02
693	High Carbon Na	22,000	317	6.5	17	0.01
694	High Carbon Na	20,000	538	7.0	14	0.0065

* All Exposures were simultaneous, therefore, exposed specimens have identical histories.

(1) Failed at gage mark; specimen data voided.

TABLE 3.10 — CREEP STRENGTH FOR
2 $\frac{1}{4}$ Cr-1 Mo STEEL IN VARIOUS
ENVIRONMENTS⁷⁹

Test Environment, 1100°F	Stress to Produce 1% Creep in 100,000 hr (psi)
Helium	5000
Air	3400
Sodium (30 ppm oxygen)	3400
Sodium (300 ppm oxygen)	3400

TABLE 3.11 — HIGH TEMPERATURE CYCLIC STRAIN (FATIGUE)
TEST† DATA FOR TYPE 316 STAINLESS STEEL⁸⁰

Exposure Environment	Exposure Time (hr)	Test Environment	Total Cyclic Strain (%)	Cycles to Failure	Average Cycles to Failure
None	0	Air	2.14	270	333
	0		2.11	360	
	0		2.14	369	
	0		1.00	1,498	1,554
	0		1.00	1,663	
	0		1.00	1,500	
	0		0.554	8,956	8,524
	0		0.568	8,060	
	0		0.565	8,556	
	0				
None	0	Helium	2.13	1,325	1,098
	0		2.19	848	
	0		2.14	1,122	
	0		1.00	9,693	8,856
	0		0.99	9,058	
	0		1.00	7,816	
	0		0.558	50,004	41,940
	0		0.571	38,804	
	0		0.573	37,011	
	0				
None	0	Low-oxygen Na	2.16	615	464
	0		2.18	275	
	0		2.16	492	
	0		2.16	475	(800)*
	0		0.985	2,385	(3,130)
	0		1.00	2,110	
	0		0.995	3,914	
	0		1.01	4,112	(4,750)*
	0		0.565	55,925	39,460
	0		0.556	29,400	
	0		0.558	33,055	
Low-oxygen Na	286	Air	2.11	296	308
	286		2.12	302	
	286		2.12	326	
	286		2.13	1,116	1,033
	286		2.16	1,233	
	286		2.11	749	
	286		2.18	700	665
	286		2.16	751	
	286		2.16	545	
	286				
Low-oxygen Na	0	High-oxygen Na	2.18	207	341
	0		2.19	427	
	0		2.18	390	
	0		1.01	4,013	(750)*
	0		1.01	5,280	
	0		1.01	3,200	
	0		0.98	4,220	4,178
	0		0.565	14,580	
	0		0.566	13,400	
	0				
None	0	High-oxygen Na	2.18	207	341
	0		2.19	427	
	0		2.18	390	
	0		1.01	4,013	(750)*
	0		1.01	5,280	
	0		1.01	3,200	
	0		0.98	4,220	4,178
	0		0.565	14,580	
	0		0.566	13,400	
	0				

* Value after correction for carbon.

† Test consists of cyclically bending a flat specimen over fixed radii at 180 cycles/hr in sodium.

TABLE 3.11 — (CONTINUED)

Exposure Environment	Exposure Time (hr)	Test Environment	Cyclic Strain (%)	Cycles to Failure	Average Cycles to Failure
None	0	High-oxygen Na	0.566	9,474	11,772
None	0	High-oxygen Na	0.554	9,633	
High-oxygen Na ↓	312	Air	2.19	404	303
	312	Air	2.18	225	
	312	Air	2.18	280	
	312	Helium	2.19	597	
	312	Helium	2.18	555	491
	312	Helium	2.19	374	
	312	Helium	2.18	439	
	312	High-oxygen Na	2.18	614	716
	312	High-oxygen Na	2.18	807	
	312	High-oxygen Na	2.18	729	
High-oxygen Na ↓	0	High-carbon Na ↑	2.18	399	306
	0		2.11	181	
	0		2.14	331	
	0		1.01	1,323	1,675
	0		1.01	1,690	
	0		0.99	2,013	
	0		0.558	12,600	15,365
	0		0.558	17,460	
	0		0.562	14,580	
	0	High-carbon Na	0.567	16,820	171
None ↓ High-carbon Na	272	Air	2.14	245	
	272	Air	2.14	144	
	272	Air	2.14	124	
	272	Helium	2.18	290	
	272	Helium	2.14	195	266
High-carbon Na ↓	272	Helium	2.14	315	
	272	High-carbon Na	2.12	103	108
	272	High-carbon Na	2.16	122	
High-carbon Na	272	High-carbon Na	2.14	100	

TABLE 3.12 — HIGH TEMPERATURE CYCLIC STRAIN (FATIGUE) TEST
DATA FOR 304 SS IN AIR, HELIUM, AND HIGH-OXYGEN SODIUM¹⁰⁰

<u>Specimen</u>	<u>Exposure Media</u> 1200 F - 287 Hrs	<u>% Total</u> <u>Cyclic</u> <u>Strain</u>	<u>Cycles to</u> <u>Failure</u>	<u>Average Cycles</u> <u>to Failure</u>	<u>Standard</u> <u>Deviation</u>	<u>Carbon⁽¹⁾</u> <u>Increase, ppm</u>
<u>Tested in High Oxygen Sodium</u>						
650	None	1.89	210			882
654	None	1.88	247			886
656	None	1.87	277	517		931
884	None	1.90	777	(1090) (2)		277
888	None	1.89	787			358
889	None	1.89	804			370
660	None	0.88	3075			472
853	None	0.88	1980	2585	577	583
854	None	0.89	3081	(2740) (2)		890
856	None	0.89	2205			378
858	None	0.50	6555			514
871	None	0.49	6724	6116	910	274
875	None	0.49	5070			155
863	High Oxygen Na	1.92	369			961
865	High Oxygen Na	1.92	363	370	8.1	1231
868	High Oxygen Na	1.90	379	(405) (2)		1121
<u>Tested in Air</u>						
641	None	1.86	445			--
642	None	1.88	525	478	42	--
644	None	1.87	465			--
645	None	0.86	2400			--
646	None	0.84	2684	2652	237	--
651	None	0.87	2871			--
655	None	0.49	29610			--
658	None	0.49	27795	29210	1263	--
652	None	0.50	30225			--
859	High Oxygen Na	1.87	285			699
860	High Oxygen Na	1.92	300	325	57	803
861	High Oxygen Na	1.90	390	(560) (2)		708
<u>Tested in Helium</u>						
902	None	1.92	900			--
903	None	1.92	1365	1144	233	--
904	None	1.92	1320			--
921	None	1.92	990			--
897	None	0.89	18495			--
899	None	0.89	15465	16370	1847	--
920	None	0.89	15150			--
894	None	0.50	98325			--
895	None	0.50	67410	76512	18976	--
922	None	0.50	63800			--
862	High Oxygen Na	1.82	774			705
864	High Oxygen Na	1.92	480	593	135	816
866	High Oxygen Na	1.92	500	(1340) (2)		827
867	High Oxygen Na	1.89	618			550

(1) Original carbon content - 480 ppm.

(2) Theoretical value after corrected for carbon increase.

TABLE 3.13 — HIGH TEMPERATURE CYCLIC STRAIN (FATIGUE) TEST DATA
FOR 304 SS IN AIR, HELIUM, AND HIGH-CARBON SODIUM AT 1200°F¹⁰⁰

<u>Specimen</u>	<u>Exposure Media</u> <u>1200 F - 427 Hrs</u>	<u>% Cyclic</u> <u>Strain</u>	<u>Cycles to</u> <u>Failure</u>	<u>Average Cycles</u> <u>to Failure</u>	<u>Standard</u> <u>Deviation</u>	<u>Carbon⁽¹⁾</u> <u>Increase, ppm</u>
<u>Tested in Air</u>						
641	None	1.86	445			--
642	None	1.88	525	478	42	--
644	None	1.87	465			--
645	None	0.86	2400			--
646	None	0.84	2684	2652	237	--
651	None	0.87	2871			--
655	None	0.49	29610			--
658	None	0.49	27795	29210	1263	--
652	None	0.50	30225			--
840	High Carbon Na	1.92	180			2642
841	High Carbon Na	1.92	587			2634
842	High Carbon Na	1.92	585	403	214	2871
843	High Carbon Na	1.92	258			2744
<u>Tested in Helium</u>						
902	None	1.92	900			--
903	None	1.92	1365	1144	233	--
904	None	1.92	1320			--
921	None	1.92	990			--
897	None	0.89	18495			--
899	None	0.89	15465	16370	1847	--
920	None	0.89	15150			--
894	None	0.50	98325			--
895	None	0.50	67410	76512	18976	--
922	None	0.50	63800			--
844	High Carbon Na	1.92	465			3081
845	High Carbon Na	1.89	531	473	55	3043
846	High Carbon Na	1.89	422			2804
<u>Tested in High Carbon Sodium</u>						
643	None	1.89	330			438
648	None	1.89	285	315	26	492
652	None	1.87	330			211
879	None	0.89	2140			1430
883	None	0.88	3220	2572	572	659
886	None	0.87	2355			428
661	None	0.49	70650			2257
857	None	0.50	74790	71497	2962	3237
872	None	0.49	69050			2156
847	High Carbon Na	1.89	117			3151
848	High Carbon Na	1.91	102	130	36	3619
849	High Carbon Na	1.92	170			4344

(1) Original carbon content - 480 ppm.

Fig. 3.1 — Compositional Changes for Type 316 SS,
Measured by Microprobe Analysis, Associated with
Elemental Leaching in Sodium⁶⁹

Fig. 3.2 — Mass Transfer in a Simulated LMFBR Primary System¹³⁹

Fig. 3.3 — Carbon Distribution in Austenitic Stainless Steel after 30-yr Exposure to 1200°F Sodium

Fig. 3.4 — Mean Carbon Content in 304 and 316 SS Monometallic System Hot Leg Components after 30 yr

Fig. 3.5 — Mean Carbon Content in $2\frac{1}{4}$ Cr-1 Mo Steel Bimetallic System Components during 30 yr

MATERIALS

- TYPE 316 STAINLESS STEEL, ANNEALED
- ▣ 316 STAINLESS STEEL, ANNEALED, 20% R OF A
- 316 L STAINLESS STEEL, ANNEALED
- △ 18/8/TI STAINLESS STEEL, ANNEALED
- △ NIMONIC P.E.16
- ◇ 2 1/4 Cr 1 Mo FERRITIC STEEL, ANNEALED, NORMALISED AND TEMPERED + 30% R OF A
- 2 1/4 Cr 1 Mo Nb Ti FERRITIC STEEL, ANNEALED, NORMALISED AND TEMPERED
- ◇ 9 Cr 1 Mo V FERRITIC STEEL

Fig. 3.6 — The Effect of Temperature and Oxygen Concentration on the Rate of Metal Loss from Various Materials after Exposure to Flowing Sodium⁶⁴

Fig. 3.7 — Effect of Temperature and Oxygen on Corrosion Rate of Various Austenitic Steels in Flowing Sodium

Fig. 3.8 — The Effect of Changing Velocity and Oxygen Concentration of Sodium on the Rate of Metal Loss from 316 Stainless Steel. Temperature: 725°C (1337°F). Specimens in Situ during the Changes.⁶⁴ (See Fig. 3.6 for Symbol Legend.)

Fig. 3.9 — The Effect of Changing the Oxygen Level of Liquid Sodium on the Rate of Metal Loss of 316 SS Specimens⁶⁹

Fig. 3.10 — Comparison of UKAEA and GE Data on Corrosion Rate of 316 SS in Sodium Containing Different Concentrations of Oxygen⁶⁶

Fig. 3.11 — Comparison of Corrosion Rates of Stainless Steels vs Reciprocal Temperature

Fig. 3.12 — Effect of Temperature and O₂ Level on the Corrosion Rate of Type 316 Stainless Steels¹³⁹

Fig. 3.13 — Loss of Biaxial Stress-Rupture Strength of Type 304 Stainless Steel in Decarburizing 1400 °F Sodium³¹

Fig. 3.14 — Metallographically Determined Depth of Volume Carburization of Type 304 SS in Carbon-Saturated Sodium vs Time at Temperature^{50, 70}

Fig. 3.15 — Decarburization of 2 1/4 Cr-1 Mo Steel Samples, 1/16-in. Thick — Exposure Time: 700 hr^{69, 79}

Fig. 3.16 — Decarburization of 2 1/4 Cr-1 Mo Steel Specimens, 1/16-in. Thick, in Sodium⁶⁹

Fig. 3.17 — Decarburization of 2 1/4 Cr-1 Mo Steel in 1100°F Sodium^{69, 79, 80}

Fig. 3.18 — Decarburization of Modified Chrome-Moly Steels in 1200°F Dynamic Sodium⁶⁹

Fig. 3.19 — 0.2% Offset Yield Strength of Candidate Ferritic Alloys — As-Received Material and Material Exposed 1000, 3000, and 4000 hr in Helium and Sodium at 1200°F — 60-mil Specimens (Phase II, Task 1)

Fig. 3.20 — Tensile Strength of Candidate Ferritic Alloys — As-Received Material and Material Exposed 1000, 3000, and 4000 hr in Helium and Sodium at 1200°F — 60-mil Specimens (Phase II, Task 1)

Fig. 3.21 — Ductility of Candidate Ferritic Alloys — As-Received Material and Material Exposed 1000, 3000, and 4000 hr in Helium and Sodium at 1200°F — 60-mil Specimens (Phase II, Task 1)

Fig. 3.22 - Carbon Variance after 1000, 3000, and 4000 hr at Temperature in 1200°F Retort Sodium (Static Conditions)

Fig. 3.23 — Effect of Velocity on Rate of Corrosion of Austenitic Stainless Steels by Flowing Sodium at 650°C (1202°F)

Fig. 3.24 — Schematic Representation of the Temperature and Oxygen Effects on the Corrosion Rates of Fe, Cr, and Ni from Stainless Steels¹³⁹

Fig. 3.25 — Corrosion Rate at 1300°F as a Function of Sodium Velocity at an Oxygen Level of Less Than 10 ppm. The influence of Downstream (Saturation) upon This Correlation also Is Shown.¹³⁹

Fig. 3.26 — Effect of Velocity on the Corrosion Rate of Type 316 Stainless Steel¹³⁹

① AND ② VELOCITY DEPENDENT REACTION

③ AND ④ VELOCITY INDEPENDENT REACTION

① VELOCITY DEPENDENT REACTION

② AND ③ VELOCITY INDEPENDENT REACTION

Fig. 3.27 — Mechanisms Proposed for the Process of Corrosion in Sodium at 1300 °F¹³⁹

Fig. 3.28 — Comparative Mass Transfer Rates of Nickel-Base Alloys and Stainless Steels in Sodium. Operating Conditions: $\Delta T - 300^{\circ}\text{F}$; flow - 2.5 gal/min; test time - 1000 hr.⁹⁹

Fig. 3.29 — Effect of ΔT on Mass Transfer in Sodium, Inconel Pump Loops^{69,98}

Fig. 3.30 — Effect of Exposure Time on Attack of Stainless Steels by Flowing Sodium

Fig. 3.31 — A Comparison of the Behavior of Various Steels in Flowing Liquid Sodium at 650°C; O₂ Level ~25 ppm; Central Specimens Taken for Comparison Purposes.⁷

Fig. 3.32 — Surface Chromium Content vs Exposure Time for Hot Leg Samples⁷

Fig. 3.33 — Surface Chromium Content vs Exposure Time for Cold Leg Samples⁷

Fig. 3.34 — Carbon Transport as a Function of the Mass Transfer of Chromium¹³⁹

○ - From ASM Data - Air; □ - From MSA Data - Air, He

Fig. 3.35 — Comparison of MSAR Tensile Data from Specimens Exposed 4000 hr in 1200°F Sodium with Other 316 Stainless Steel Heats¹³³
(Data from Metals Handbook, Volume 1 — Properties and Selection of Metals, American Society for Metals)

Fig. 3.36 — Tensile Strength — Type 316
Stainless Steel⁸⁰

Fig. 3.37 — Yield Strength — Type 316
Stainless Steel^{80,102}

Fig. 3.38 — 25% Tensile Strength —
Type 316 Stainless Steel^{80,102}

EXPOSURE MEDIA 4000 hr AT 1200°F	TEST MEDIA AND TEST TEMPERATURE
○ NONE	AIR-1200°F
□ NONE	He-RT AND 1200°F
△ Na - LOW OXYGEN	He-RT AND 1200°F
◊ Na - CARBON	AIR-RT AND He-1200°F
▲ He	AIR-RT AND He-1200°F
◡ Na - HIGH OXYGEN	AIR-RT AND He-1200°F
● He	AIR-RT AND He-1200°F

NOTE:
POINTS ARE 62 1/2 % OF YIELD DATA

Fig. 3.39 — 62 1/2% Yield Strength —
Type 316 Stainless Steel^{80, 102}

Fig. 3.40 — Elongation — Type 316
Stainless Steel^{80, 102}

EXPOSURE MEDIA 4000 hr AT 1200°F	TEST MEDIA AND TEST TEMPERATURE
○ NONE	AIR-1200°F
□ NONE	He-RT AND 1200°F
△ Na - LOW OXYGEN	He-RT AND 1200°F
◊ Na - CARBON	AIR-RT AND He-1200°F
▲ He	AIR-RT AND He-1200°F
◊ Na - HIGH OXYGEN	AIR-RT AND He-1200°F
● He	AIR-RT AND He-1200°F

Fig. 3.41 — Effect of Carbon Content on Short-Time Yield Strength (0.2% Offset) of Type 304 Stainless Steel at Room Temperature and 1200°F^{27,104}

Fig. 3.42 — Effect of Carbon Content on Elongation at Fracture in Short-Time Tensile Tests of Type 304 Stainless Steel at Room Temperature and at 1200°F^{27,104}

ROOM TEMPERATURE TESTS

1200 °F TESTS IN AIR

ROOM TEMPERATURE TESTS

1200°F TESTS IN AIR

(UW) SPECIMENS NOT WASHED PRIOR TO TESTING - ALL OTHERS WASHED WITH ALCOHOL AND WATER
ALL EXPOSURES (4000 HOURS) SIMULTANEOUS, THEREFORE SPECIMENS HAVE IDENTICAL THERMAL HISTORIES

Fig. 3.43 — Tensile Strength of 304SS (High Oxygen Sodium)¹⁰⁵

ROOM TEMPERATURE TESTS

1200°F TESTS IN AIR

ROOM TEMPERATURE TESTS

1200°F TESTS IN AIR

(UW) SPECIMENS NOT WASHED PRIOR TO TESTING - ALL OTHERS WASHED WITH ALCOHOL AND WATER
ALL EXPOSURES (4038 HOURS) SIMULTANEOUS, THEREFORE SPECIMENS HAVE IDENTICAL THERMAL HISTORIES

Fig. 3.44 — Tensile Test Data of 304 SS after Exposure to 1200°F Helium and High Carbon Sodium¹⁰⁵

EXPOSURE MEDIA 4000 hr AT 1100°F		TEST MEDIA AND TEST TEMPERATURE
○	NONE	AIR AT RT AND 1100°F
□	NONE	He AT 1100°F
△	Na - LOW OXYGEN (30 ppm)	He AT 1100°F
◇	Na - HIGH OXYGEN (300 ppm)	He AT RT AND 1100°F
x	HELIUM	He AT RT AND 1100°F

NOTE:
POINTS ARE 25% OF MEASURED TENSILE
VALUES

Fig. 3.45 — 25% Tensile Strength of $2\frac{1}{4}$ Cr-1 Mo Steel⁸⁰

EXPOSURE MEDIA 4000 hr AT 1100°F		TEST MEDIA AND TEST TEMPERATURE
○	NONE	AIR AT RT AND 1100°F
□	NONE	He AT 1100°F
△	Na - LOW OXYGEN (30 ppm)	He AT 1100°F
◇	Na - HIGH OXYGEN (300 ppm)	He AT RT AND 1100°F
x	HELIUM	He AT RT AND 1100°F

NOTE:
POINTS ARE 62 1/2 % OF MEASURED YIELD
VALUES

Fig. 3.46 — 62 1/2% Yield Strength of $2\frac{1}{4}$ Cr-1 Mo Steel⁸⁰

Fig. 3.47 — Minimum Creep Rate of Type 316 SS in Air, Helium, and High- and Low-Oxygen Sodium at 1200°F⁸⁰

Fig. 3.49 — Creep Rate — Type 316 Stainless Steel⁸⁰ (1% in 100,000 hr)

Fig. 3.48 — Minimum Creep Rate of Type 316 SS in Air, Helium, and High-Carbon and Low-Oxygen Sodium at 1200°F⁸⁰

EXTRAPOLATED RUPTURE DATA		
CURVE	EXPOSURE MEDIA 4000 hr AT 1200°F	TEST MEDIA AT 1200°F
1	Na - CARBON (30-50 ppm)	Na - CARBON (30-50 ppm)
2	NONE	Na - CARBON (30-50 ppm)
3	NONE	AIR HELIUM Na - LOW OXYGEN (30 ppm OXYGEN) Na - HIGH OXYGEN (300 ppm OXYGEN)

Fig. 3.50 — Creep-Rupture Strength — Type 316 Stainless Steel — Extrapolated to 100,000 hr⁸⁰

EXTRAPOLATED 100,000 hr RUPTURE DATA	
EXPOSURE MEDIA 4000 hr AT 1200°F	TEST MEDIA AT 1200°F
<div> <div> <div>NONE</div> <div>NONE</div> <div>NONE</div> <div>NONE</div> <div>Na - LOW OXYGEN</div> <div>Na - LOW OXYGEN</div> <div>Na - LOW OXYGEN</div> <div>Na - HIGH OXYGEN</div> <div>Na - HIGH OXYGEN</div> <div>Na - HIGH OXYGEN</div> </div> <div> <div>△</div> <div>○</div> <div>□</div> </div> </div>	<div> <div>AIR</div> <div>HELIUM</div> <div>Na - LOW OXYGEN</div> <div>Na - HIGH OXYGEN</div> <div>AIR</div> <div>HELIUM</div> <div>Na - LOW OXYGEN</div> <div>AIR</div> <div>HELIUM</div> <div>Na - HIGH OXYGEN</div> </div> <div> <div>Na - CARBON</div> <div>Na - CARBON</div> </div>

Fig. 3.51 — 100,000-hr Creep-Rupture Strength — Type 316 Stainless Steel⁸⁰

Fig. 3.52 — Design Curves — Type 316 Stainless Steel in High-Oxygen Sodium at 1200°F⁸⁰

Fig. 3.53 — Design Curves — Type 316 Stainless Steel Exposed in High-Oxygen Sodium at 1200°F⁸⁰

NOTE:
ALL MATERIAL EXPOSED 4000 hr AT
1200°F IN HIGH-OXYGEN SODIUM (300 ppm
O₂) AND THEN TESTED IN SAME MEDIA
 $\% = \text{STRAIN } \left(\frac{\Delta L}{L} \right)$

Fig. 3.54 — Design Curves — Type 316 Stainless Steel in High-Carbon Sodium at 1200°F⁸⁰

NOTE:
ALL MATERIAL EXPOSED 4000 hr AT
1200°F IN HIGH-CARBON SODIUM
AND THEN TESTED IN SAME MEDIA
 $\% = \text{STRAIN } \left(\frac{\Delta L}{L} \right)$

Fig. 3.55 — Design Curves — Type 316 Stainless Steel Exposed in High-Carbon Sodium at 1200°F⁸⁰

Fig. 3.56 — Rupture Strength of 304 SS (Original Material) in High-Oxygen Sodium at 1200°F¹⁰⁰

Fig. 3.57 — Comparison of Rupture Strength of 304 SS (Original Material and 4000-hr Exposed Material) in Air and High-Oxygen Sodium at 1200°F¹⁰⁰

Fig. 3.58 — Comparison of Creep Strength of 304 SS (Original Material and 4000-hr Exposed Material) in Air and High-Oxygen Sodium at 1200 °F¹⁰⁰

Fig. 3.59 — Biaxial Stress-Rupture Properties of Type 304 Stainless Steel in High-Purity Sodium¹¹⁰

Fig. 3.60 — Biaxial Stress-Rupture Properties of Type 316 Stainless Steel in High-Purity Sodium¹¹⁰

Fig. 3.61 — Time-Exposure Representation of X8CrNiMoVNB1613 Steel at 700°C¹⁸⁴

Fig. 3.62 — Time-Exposure Representation of Incoloy-800 at 700°C with and without the Presence of Sodium¹⁸⁴

Fig. 3.63 — Secondary Creep Speed of Incoloy-800 and X8CrNiMoVNB1613 Steel Depending on Tension at 700°C¹⁸⁴

Fig. 3.64 — Stress-Rupture of X8CrNiMoVNB1613 Steel in Time-Exposure Test at 700°C with and without the Presence of Sodium¹⁸⁴

Fig. 3.65 — Stress-Rupture of Incoloy-800 Weld in Time-Exposure Test at 700°C with and without the Presence of Sodium¹⁸⁴

Fig. 3.66 — Creep Curves of X8CrNiMoVNB1613 Steel at 18 kp/mm² and 700°C¹⁸⁴

Fig. 3.67 — Minimum Creep Rate of $2\frac{1}{4}$ Cr-1 Mo Steel at 1100°F^{79,80}

Fig. 3.68 — Creep-Rupture Strength
of $2\frac{1}{4}$ Cr-1 Mo Steel before 4000-hr
Exposure^{79,80}

Fig. 3.69 — Creep-Rupture Strength
of $2\frac{1}{4}$ Cr-1 Mo Steel after 4000-hr
Exposure^{79,80}

Fig. 3.70 — 100,000-hr Creep-Rupture
Strength of $2\frac{1}{4}$ Cr-1 Mo Steel^{79,80}

Fig. 3.71 — Design Curves — $2\frac{1}{4}$ Cr-1 Mo Steel in High-Oxygen Sodium at 1100°F ^{79,80}

Fig. 3.72 — Design Curves — $2\frac{1}{4}$ Cr-1 Mo Steel Exposed in High-Oxygen Sodium at 1100°F ^{79,80}

Fig. 3.73 — Impact Test Results for Type 316 SS
Following Exposure in Various Environments⁸⁰

Fig. 3.74 — 304 SS Charpy Keyhole Impact Strength (Original Material and Material Exposed to High-Oxygen Sodium and Helium at 1200°F). Specimen: 0.394-in. wide × 0.098-in. thick.¹⁰⁰

Fig. 3.75 — 304 SS Charpy Keyhole Impact Strength — Material Exposed to High-Carbon Sodium at 1200 °F¹⁰⁰ (Test 7 — Specimen: 0.384-in. wide × 0.098-in. thick)

Fig. 3.76 — Room Temperature Charpy Keyhole Impact Test Results for $2\frac{1}{4}$ Cr-1 Mo Steel after 1200°F Exposure^{79,80}

Fig. 3.77 — Fatigue Behavior — Type 316 SS
in Air, Helium, Low-Oxygen Sodium, High-
Oxygen Sodium, and High-Carbon Sodium
Environments at 1200°F⁸⁰

Fig. 3.78 — Plastic Strain vs Cycles to Failure — Type 316 SS
at 1200°F⁸⁰

Fig. 3.79 — High-Frequency Fatigue of Type 316 SS in 1000°F Liquid Sodium at Two Oxide Levels¹⁰⁹

Fig. 3.80 — High-Frequency Fatigue of 316 SS in 1500°F Sodium at Two Oxygen Levels¹⁰⁹

Fig. 3.81 — Comparison of 304 SS (Original Material and 287-hr Exposed Material) Cycle Life in Air, Helium, and High-Oxygen Sodium at 1200 F¹⁰⁰

Fig. 3.82 — Comparison of 304 SS (Original Material and 427-hr Exposed Material)
Cycle Life in Air, Helium, and High-Carbon Sodium at 1200°F¹⁰⁰

Fig. 3.83 — Fatigue Tests — $2\frac{1}{4}$ Cr-1 Mo Steel in Sodium, Air, and Helium^{79,80}

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APPENDIX

CASE
1423-1

INTERPRETATIONS OF ASME BOILER AND PRESSURE VESSEL CODE

Approved by Council October 31, 1969

Case 1423-1
(Special Ruling)
Wrought Type 304 & 316 with Nitrogen
Added.
Sections I, III, VIII, Divisions 1 and 2

Inquiry: Is it permissible in welded construction to the requirements of Section I, III, and VIII to use the austenitic grades of 304 and 316, with controlled additions of nitrogen, as

wrought pipes and tube and as other wrought product forms conforming to the requirements of the applicable SA material specification listed in Tables PG-23.1 of Section I, N-421 of Section III, UHA-23 of Section VIII Division 1, and AHA-1 of Section VIII Division 2, except that the material shall conform to the following chemical and tensile requirements?

1. The chemical requirements shall be as follows:

	<u>Grade 304N</u>	<u>Grade 316N</u>
Carbon, % max.	0.08	0.08
Manganese, % max.	2.00	2.00
Phosphorus, % max.	0.03	0.03
Sulphur, % max.	0.03	0.03
Silicon, % max.	0.75	0.75
Nickel, %	8.0 to 11.0	11.0 to 14.0
Chromium, %	18.0 to 20.0	16.0 to 18.0
Molybdenum, %	—	2.0 to 3.0
Nitrogen, %	0.10 to 0.16	0.10 to 0.16

2. The room temperature minimum mechanical properties shall be as follows:

	<u>Grade 304N</u>	<u>Grade 316N</u>
Ultimate Tensile Strength, psi	80,000	80,000
.2% Offset Yield Strength, psi	35,000	35,000
% Elongation—2" Gage Length		
Longitudinal	30	30
Transverse	25	25
% Reduction of Area		
Longitudinal	50	50
Transverse	45	45

Meeting of September 12, 1969

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CASE (continued) 1423-1

INTERPRETATIONS OF ASME BOILER AND PRESSURE VESSEL CODE

Reply: It is the opinion of the Committee that the material described in the Inquiry may be used under the rules of Sections I, III, and VIII sub-

ject to the following additional requirements:

1. The design stresses shall be as follows:

Temp. (°F)	Design Stress, KSI			
	Maximum Allowable Stress		Stress Intensity	
	Section I and Section VIII, Division 1		Section III and Section VIII, Division 2	
	Grade 304N	Grade 316N	Grade 304N	Grade 316N
75	20.0	20.0	20.0	20.0
100	20.0	20.0	20.0	20.0
200	17.9	20.0*	19.4	20.0*
300	15.7	19.0*	17.8	19.2*
400	14.1	18.3*	16.5	18.8*
500	13.0	17.8*	15.4	18.6*
600	12.4	17.4*	14.6	18.6*
650	12.2	17.3*	14.2	18.6*
700	11.9	17.15*	13.9	18.6*
750	11.75	16.9*	13.6	18.5*
800	11.55	16.65*	13.3	18.4*
850	11.3	16.3*	13.1	18.3*
900	11.05	15.9*	12.8	18.1*
950	10.8	15.6*	12.6	17.8*
1000	10.55	15.0*	12.4	17.4*
1050	10.3	12.4*	12.2	15.8*
1100	9.75	9.75	11.7	12.4*
1150	7.7	7.7	9.8	9.8
1200	6.05	6.05	7.4	7.4

*Due to the relatively low yield strength of these materials, these higher stress values were established at temperatures where the short time tensile properties govern to permit the use of these alloys where slightly greater deformation is acceptable. The stress values in this range exceed 62-1/2 per cent but do not exceed 90 per cent of the yield strength at temperature. Use of these stresses may result in dimensional changes due to permanent strain.

CASE (continued) 1423-1

INTERPRETATIONS OF ASME BOILER AND PRESSURE VESSEL CODE

2. For welded tube or pipe to SA-249 or SA-312 used under the rules of Section I or Section VIII Divisions 1 and 2, the design stress listed above shall be multiplied by a joint efficiency factor of 0.85.

The design fatigue curve of Fig. N-415 (B) of Section III, and Fig. 5-110.2 of Section VIII, Div. 2 shall apply.

The material shall receive P-Number 8 classification in Table Q-11.1 of Section IX.

Yield strength values shall be as follows:

Table 2

Minimum Yield Strength Values, Ksi

Temp. °F	304N	316N
75	35.0	35.0
100	33.6	35.0
200	28.7	31.0
300	25.1	28.5
400	22.6	26.4
500	20.8	24.7
600	19.8	23.4
650	19.5	22.8
700	19.1	22.3
750	18.8	21.8
800	18.5	21.3
850	18.1	20.9
900	17.7	20.5
950	17.3	20.2
1000	16.9	19.8

TABLE A.4 — MINIMUM YIELD
STRENGTH VALUES, KSI

Temp, °F	304N	316N
75	35.0	35.0
100	33.6	35.0
200	28.7	31.0
300	25.1	28.5
400	22.6	26.4
500	20.8	24.7
600	19.8	23.4
650	19.5	22.8
700	19.1	22.3
750	18.8	21.8
800	18.5	21.3
850	18.1	20.9
900	17.7	20.5
950	17.3	20.2
1000	16.9	19.8