

# Clinch River Breeder Reactor Plant Nuclear Island

## SELECTION OF COOLANT-BOUNDARY MATERIALS FOR THE CLINCH RIVER BREEDER REACTOR PLANT

AUGUST 1974

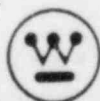
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SELECTION OF COOLANT-BOUNDARY MATERIALS FOR  
THE CLINCH RIVER BREEDER REACTOR PLANT

P. 500

APPROVED:



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## SUMMARY

A study has been made of the pertinent material properties which will influence the selection of a particular grade of austenitic stainless steel in primary and intermediate system CRBRP components. Types 304 and 316 stainless steel were considered, together with their low and high-carbon derivations. General guidelines were established which will enable the optimum choice of material to be made. A description of the properties of ferritic 2-1/4 Cr-1 Mo steel is also presented in order to define areas where it could possibly be used in place of stainless steel. It should be noted that 2-1/4 Cr-1 Mo steel has already been selected as the steam generator reference alloy.

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## 1. INTRODUCTION

A study has been made of the properties of Types 304 and 316 stainless steel in order to critically assess their relative merits for coolant-boundary applications in liquid metal fast breeder reactor (LMFBR) systems. Materials in the low, normal and high carbon grades are considered. The objective of this evaluation is to make recommendations for the selection of materials for components in the primary, intermediate and auxiliary systems of the CRBRP based on material performance and, to a lesser extent, cost. For the steam generator an additional evaluation has been made of the properties of 2-1/4 Cr-1 Mo low alloy steel which will be used in this component.

The components of interest include, but are not necessarily limited to, the following:

### Primary System

- a) Vessel
- b) Vessel outlet nozzle
- c) Vessel inlet nozzle
- d) Hot-leg piping
- e) Cold-leg piping
- f) Pump suction nozzle
- g) Pump discharge nozzle
- h) Pump casing
- i) Intermediate heat exchanger inlet nozzle
- j) Intermediate heat exchanger outlet nozzle
- k) Intermediate heat exchanger shell
- l) Hot-leg isolation valve inlet nozzle
- m) Hot-leg isolation valve outlet nozzle
- n) Valve body
- o) Cold-leg isolation valve inlet nozzle
- p) Cold-leg isolation valve outlet nozzle
- q) Cold-leg isolation valve body

#### Intermediate System

- a) Intermediate heat exchanger inlet nozzle
- b) Intermediate heat exchanger outlet nozzle
- c) Hot-leg piping
- d) Cold-leg piping
- e) Pump suction nozzle
- f) Pump discharge nozzle
- g) Pump casing
- h) Hot-leg isolation valve
- i) Hot-leg isolation valve outlet nozzle
- j) Hot-leg isolation valve body
- k) Cold-leg isolation valve inlet nozzle
- l) Cold-leg isolation valve outlet nozzle
- m) Cold-leg isolation valve body
- n) Steam generator tubing
- o) Steam generator inlet nozzle
- p) Steam generator outlet nozzle
- q) Steam generator shell

#### Auxiliary System

- a) Auxiliary system hot-leg piping
- b) Auxiliary system hot-leg sample loop piping

For a particular application it is not always possible to differentiate between the behavior of Type 304 and Type 316 stainless steel. For example, in sensitization, stress-corrosion cracking, fabrication and weldability the differences between the two materials are small. The selection of one over the other is, therefore, often made on a lowest-cost basis. More importantly, in this evaluation, significant emphasis is placed on determining the merits of alloys containing various starting contents of carbon and nitrogen since it is vital to estimate the long-term effects of interstitial transfer on mechanical behavior in an LMFBR system.

Property evaluations for the various materials are given in the following sections.

## 2. PROPERTIES OF TYPES 304 AND 316 STAINLESS STEEL

### 2.1 Mechanical Properties

#### 2.1.1 Short-Term Tensile Properties

The mechanical properties of the unstabilized austenitic stainless steels are influenced by the presence of interstitial elements such as carbon and nitrogen. Although it has been demonstrated that nitrogen has as strong an effect as carbon,<sup>(1)</sup> and most Type 304 and Type 316 stainless steel contain more nitrogen than carbon (Figure 1), the ASME Boiler and Pressure Vessel Code does not address the problem of nitrogen content in stainless steel. Classifications within a particular grade of steel are given in terms of carbon content only (see Table 1). For high temperature service ( $> 800^{\circ}\text{F}$ ) low-carbon grades of material are not permissible and the minimum carbon content for normal grades is specified as 0.04 weight percent.<sup>(3)</sup> Depending on the carbon content of the material, different allowable design stresses may be specified in the ASME Code. It is, therefore, important to assess whether the interstitial content in a sodium-exposed reactor component falls below specified minimums during service and also to determine the impact of changing interstitial content on mechanical properties.

##### 2.1.1.1 Effect of Carbon and Nitrogen

A comprehensive analysis has been completed on the effects of carbon and nitrogen concentrations on the yield strength, tensile strength and total elongation of solution-treated Types 304 and 316 stainless steel.<sup>(1)</sup> Since most of the available data in the literature do not include nitrogen contents, the amount of useful information available is small, especially for Type 316.

The following assumptions and guidelines were used in selecting data for analysis in reference 1:

- a) All alloys must have chemical compositions within the limits specified by the American Iron and Steel Institute (AISI).
- b) Plate, bar, pipe, tube, and forged materials were considered.
- c) Carbon and nitrogen were assumed to have equal and additive effects on mechanical behavior; hence, the total interstitial content may be given as  $(C + N)$ . For convenience, C and N are given in weight percent although it is more correct to use atom percent.
- d) Alloy grain size was assumed to be a negligible variable in the analysis.
- e) The test strain rate was assumed to be a negligible variable in the analysis.
- f) Only solution-treated material was considered. The solution temperatures were between 1900 and 2100°F, with most alloys being water or air quenched.
- g) In developing the design equations the mathematical form were generated, where possible, from existing mechanical behavior theories. This increased confidence in extrapolating the equations to very low  $(C + N)$  levels where no data exist. Where theory could not be utilized empirical data fits were made.

For Type 304 stainless steel the yield strength, ultimate tensile strength and total elongation data were shown to be<sup>(1)</sup> respectively:

$$\sigma_y = 89.952 + 181.167 (C + N) - 0.148T (C + N) - 5.727 T^{1/2} + 0.105T \quad (1)$$

standard deviation = 2.236

[Valid for  $297 < T < 922^\circ\text{K}$ ,  $75\text{-}1200^\circ\text{F}$ ]

$$\sigma_u = -1.806 - 227.391 (C + N) + 1218.392T^{-1/2} + 1.127T (C + N) - 8.446 \times 10^{-4} T^2 (C + N) \quad (2)$$

standard deviation = 2.645

[Valid for  $366 < T < 922^\circ\text{K}$ ,  $200\text{-}1200^\circ\text{F}$ ]

$$\epsilon = 29.595 - 89.898 (C + N) + 468.414T^{-1/2} \quad (3)$$

standard deviation = 2.955

[Valid for  $477 < T < 922^\circ\text{K}$ ,  $400\text{-}1200^\circ\text{F}$ ]

These equations are valid to a maximum interstitial concentration of 0.13 weight percent. The yield and tensile strengths are in ksi, the total elongation in weight percent and the temperature in degrees Kelvin.

For Type 316 stainless steel the equations are:

$$\sigma_y = 88.474 + 211.640 (C + N) - 0.206T (C + N) - 5.762T^{1/2} + 0.111T \quad (4)$$

standard deviation = 2.806

$$\sigma_u = 8.563 - 461.665 (C + N) + 1064.154T^{-1/2} + 2.722T (C + N) - 2.330 \times 10^{-3} T^2 (C + N) \quad (5)$$

standard deviation = 6.458

$$\epsilon = 28.000 - 99.139 (C + N) + 653.926T^{-1/2} \quad (6)$$

standard deviation = 6.703

The above equations are usually valid for temperatures between 297-977°K (75-1300°F) and  $(C + N) \leq 0.010$  weight percent. However, at room temperature, 1100°F and 1300°F, the range of validity may be increased, respectively, to interstitial levels of 0.167, 0.141, and 0.141 weight percent.

In Figures 2 through 5 are given direct comparisons between Type 304 and Type 316 stainless steel for various  $(C + N)$  levels. For equivalent interstitial levels there is very little difference in the yield strengths of the two alloys but the ultimate tensile strength and ductility of Type 316 stainless steel are markedly higher. However, for the lower interstitial levels the differences in the ultimate tensile strength of the two materials becomes smaller.



An important observation in Figures 3 and 4 is that a Type 304 or Type 316 steel containing a minimum required 0.04 weight percent carbon alone would not meet the specified minimum room temperature yield strength of 30 ksi. However, alloys containing an additional 0.04 weight percent nitrogen would meet this requirement. This clearly illustrates the necessity to know the nitrogen as well as the carbon content in an austenitic stainless steel.

In Table 2 is given the minimum anticipated yield strengths ( $S_y$ ) and the time-independent allowable design stresses ( $S_m$ ) for the various grades of Type 304 and Type 316 stainless steel. For normal and high carbon grades the  $S_y$  and  $S_m$  values for Type 316 stainless steel are higher. For the low-carbon grades, values for Type 304 are higher. However, if the effects of nitrogen are considered, as was the case in Figures 2 through 5, the yield strength of Type 316 stainless steel is always marginally higher. Hence, if a direct comparison is made between the tensile behavior of Types 304 and 316 stainless steel for low, normal and high interstitial grades Type 316 is superior. For any given (C + N) concentration it has higher strength than Type 304 and it has approximately 5 percent more ductility over a wide range of temperature (Figures 2 through 5).

The above discussion has centered on solution-treated material. Thermal aging causes the precipitation of carbides and intermetallic phases<sup>(4)</sup> and results in changes in mechanical behavior. Available data indicate that there is a small increase in yield strength a zero or marginal increase in tensile strength and a decrease in total elongation.<sup>(5, 6)</sup>

### 2.1.2 Creep Properties

The creep properties of Types 304 L, 304, 304 H, 316 L, 316 and 316 H stainless steels have been compared by Smith.<sup>(7)</sup> Most of the data are for temperatures of 1000°F and higher since creep at lower temperatures is extremely slow. A summary of the creep properties in the range 1000 to 1100°F is given in Figures 6 and 7. No correlation was made with nitrogen contents but it may be seen

that L grades of Type 304 and Type 316 have faster secondary creep rates than the regular carbon grades. Type 316 stainless steel has a slower creep rate than Type 304 at any given stress level but at 1100°F the creep rates of Type 304 and 316 L are similar. The rupture strengths of the L grades are less than those for the regular grades. Type 316 has a higher rupture strength than Type 304 but Type 304 and Type 316 L behave similarly.

Rupture strength data which take into account nitrogen, have been plotted for Type 304 and Type 316 solution annealed bar and are given in Figures 8-11.<sup>(8)</sup> Although both  $10^3$  and  $10^5$  hour rupture strength data are plotted the latter were obtained by extrapolation and, hence, may be subject to some error. The following general conclusions may be drawn from the data in these figures:

- a) Both carbon and nitrogen concentrations must be specified since both elements strongly influence creep behavior.
- b) Based on the small degree of scatter in the data, to a reasonable approximation the nitrogen and carbon effects are of equal magnitude so that the effective interstitial concentration may be written simply as  $(C + N)$ .
- c) Increasing  $(C + N)$  content increases the rupture strength. With the exception of the  $10^3$  hour rupture strengths for Type 316 at 1100 and 1200°F, the rupture strength is a linear function of the  $(C + N)$  content within the concentration limits evaluated.
- d) For equivalent  $(C + N)$  concentrations and test temperatures the  $10^3$  hour rupture strengths of Type 304 stainless steel are about 20-30 percent less than those for Type 316. The  $10^5$  hour rupture strengths of Type 304 are about 40 percent less than those for Type 316.

- e) The effect of (C + N) concentration on rupture strength decreases with increasing test temperature.

A limited amount of work has been conducted on the uniaxial creep of Type 316 stainless steel which was decarburized and denitrided in flowing sodium prior to testing.<sup>(9)</sup> From stress relaxation tests at 1325°F for stresses below 13 ksi the creep rates were obtained and are given in Figure 12. For the sodium-exposed material there was a loss in carbon and nitrogen from 0.0558 and 0.0556 weight percent, respectively, to 0.0466 to 0.0106 weight percent, respectively. Compared with the argon-aged material a (C + N) change from 0.1114 to 0.0572 increased the creep rate by an order of magnitude.

All of the above data clearly show the large effect which carbon and nitrogen transfer will have on the creep behavior of unstabilized austenitic stainless steel in an LMFBR environment.

### 2.1.3 Fatigue Properties

A comparison between the fatigue behavior of Type 304 and Type 316 stainless steel has been made for temperatures between 806 and 1202°F.<sup>(10)</sup> The Type 304 contained 0.053 and 0.052 weight percent carbon and nitrogen, respectively, and the Type 316 0.06 and 0.048 weight percent carbon and nitrogen, respectively. For equivalent heat treatments and test temperatures Type 316 has a shorter fatigue life than Type 304 at the same total strain range. If the comparison is based on the stress amplitude, however, Type 316 is superior. This is because for a given stress the corresponding strain range for Type 316 is somewhat smaller than that for Type 304 owing to the higher yield strength of Type 316 stainless steel. Hence, a selection of Type 304 or Type 316 for fatigue resistance will depend on whether the fatigue condition will be strain or stress controlled.

The effect of carbon content on the fatigue behavior of 18 Cr-12 Ni iron-based alloys has been studied by Driver.<sup>(11)</sup> It was shown that at elevated temperatures in the range 1112 to 1472°F, for a wide range of stress amplitudes, decreasing the carbon level from 0.05 to 0.004 weight percent causes a much shorter fatigue life and endurance limit. This was attributed to carbides effectively blocking grain-boundary sliding and migration, thereby inhibiting the nucleation of grain boundary cracks. At room temperature, however, it has been shown that for Type 316 stainless steel fatigued at strain amplitudes in the range  $\pm 1$  to  $\pm 4$  percent the presence of 2 to 3 volume percent of chromium carbide can decrease the fatigue life to about one-third of that in solution treated material.<sup>(12)</sup> This is caused by the brittle fracture of the grain-boundary carbides. There are no data for comparisons in the current temperature range of interest, viz. 650 to 1000°F, hence the effect of interstitial concentration is not known with certainty.

## 2.2 Sensitization And Heat Treatment Effects

An unstabilized austenitic stainless steel is said to be sensitized when carbides are precipitated at grain boundaries during high-temperature exposure. Precipitation of the carbides occurs readily in the temperature range 800-1600°F.<sup>(13)</sup>

When the degree of sensitization is such that continuous carbide paths are formed at the grain boundaries intergranular corrosion may result if the material is exposed to a corrosive atmosphere. A simple theory to explain this phenomenon is based on the formation of continuous  $\text{Cr}_{23}\text{C}_6$  layers at the boundaries which results in the formation of adjacent zones which are denuded of chromium.<sup>(14)</sup> In these regions therefore, a decreased resistance to corrosion occurs and intergranular attack may occur. Particular agents which have been reported to cause intergranular attack in sensitized austenitic stainless steels are given in Table 3. In the early stages of carbide precipitation, however, when continuous grain-boundary carbide paths have not been established, significant intergranular corrosion is not likely.

### 2.2.1 Measurement of Sensitization

The degree of carbide precipitation is dependent on both the time and temperature of exposure. At low temperatures ( $\sim 800^{\circ}\text{F}$ ) precipitation is slow because of the slow rate of diffusion of chromium to the grain-boundary carbides. At high temperatures ( $\sim 1600^{\circ}\text{F}$ ) the degree of supersaturation of carbon in the alloy is relatively low and there is less tendency for carbides to be formed. The tests usually employed to detect sensitization are described in ASTM procedure A262.<sup>(15)</sup> Typical sensitization curves obtained are given in Figure 13. The data represent the corrosion rates in boiling Strauss solution (acidified copper sulfate) for material pre-exposed at the temperatures and times shown. Note that for exposure temperatures of about  $1300^{\circ}\text{F}$  sensitization, manifested by increased corrosion rates, occurs at shorter exposure times. It is also clear that for the low-carbon material sensitization is significantly delayed.

### 2.2.2 Occurrence of Sensitization in LMFBR Components

Sensitization is likely to occur during fabrication, stress relieving and service of most of the components listed in Section 1 of this report. During fabrication, for example, the cooling of components after hot forming and welding will result in passing through the sensitizing temperature range. For heavy section components it may not be possible to quench rapidly enough, so sensitization is inevitable.

Stress relieving is necessary in many structures to ensure dimensional stability during service. The data given in Table 4 show the degree of stress relief which can be achieved in austenitic stainless steels for various annealing temperatures. Ideally, stress relief should not be conducted in the  $800$  to  $1600^{\circ}\text{F}$  sensitization range. If possible, it should also be avoided at high temperatures since necessary rapid cooling through the sensitization range might restore residual stresses and cause distortion.

In the case of sensitization during plant operation many of the components listed in Section 1 of this report are designed to operate over a 250,000 hour lifetime at temperature up to about 1050°F. Under such conditions sensitization is unavoidable, as Figure 13 shows.

#### 2.2.3 Effects of Sensitization

The presence of a sensitized structure is itself not a serious problem. Compared with a solution-treated material sensitization in an alloy will cause only small increases in strength and small ductility losses.<sup>(5, 6)</sup> However, as shown in Table 3 a particularly large range of corrosive agents will cause some degree of grain boundary attack in a sensitized alloy. Rusting, for example, is quite common in the heat-affected zones of welded components. Probably the most important effect of sensitization is that during fabrication, shipping or erection of LMFBR components intergranular corrosion will form incipient cracks which may act as stress raisers. This could induce stress-corrosion cracking for certain environments which the component is later exposed to. This problem is discussed in Section 2.3.

#### 2.2.4 Comparison of Alloys for Resistance to Sensitization Effects

A straightforward comparison between Type 304 and Type 316 for resistance to sensitization-induced corrosion is difficult for the following reasons:

- a) Data for alloys containing the same carbon concentration are scarce.
- b) There is difficulty in obtaining data on alloys containing typical concentrations of alloying elements, especially chromium.



- c) The presence of nitrogen could influence sensitization effects. Work reported by Hazelton indicates that nitrogen additions resulted in decreased amounts of sensitization.<sup>(20)</sup> Data which support this conclusion are given in Tables 5 and 6.<sup>(21, 22)</sup>

From fundamental considerations the higher chromium contents in Type 304 would tend to reduce the extent of chromium depleted zones at the grain boundaries. Also, molybdenum additions (Type 316 has ~ 2.5 percent molybdenum) encourages sensitization at temperatures below 1200°F<sup>(23)</sup> (Figure 14). Experimental evidence to support the contention that Type 304 stainless steel is less susceptible to sensitization is given in Figures 15 and 16 for low-carbon materials. There are no comparable data for regular grades of steel, hence it may be premature to state that Type 304 is definitely superior under all possible sensitizing conditions.

Under certain situations it may be possible to consider weaker low-carbon grades of steel to overcome sensitization problems. This may be feasible, for example, in components not subjected to the highest stresses or where operating temperatures are low. Figure 13 shows the beneficial effects of low-carbon material in delaying the onset of sensitization. Samples of Type 304 L and 316 L have been heated at 1200°F for 1 to 3 days and have shown no intergranular attack in the ASTM-262 test after three successive 72-hour immersions in the Strauss solution.<sup>(25)</sup> Hence, the use of low-carbon materials are of significant benefit since there is far greater latitude in welding and stress relieving operations conducted at high temperatures. For normal and high-carbon grades stress relief below about 950°F may be required but this will only reduce peak stresses by about 25 to 30 percent in a 2 to 4 hour annealing period.<sup>(17)</sup> If a full stress relieving treatment is mandatory an anneal above 1600°F is required (see Table 4) followed by as rapid a cool as possible through the



sensitization range. The permissible cooling rate will depend on the section thickness since care must be exercised to avoid re-introducing residual stresses and distortion.

## 2.3 Stress-Corrosion Cracking

Stress-corrosion cracking is a recurring problem for austenitic stainless steel components in the nuclear industry. It is a brittle failure phenomenon caused by the conjoint action of stress, environment and possible temperature. In references 26 and 27 a review of case histories and laboratory tests showed that the following agents were possible contributing causes of stress-corrosion: chlorides, fluorides, caustic, tar oils, natural oils, organic solvents, orange juice, milk, welding flux, ammonia, sulphates, acetic acid, alcohols, corn syrup, sulphides, baked beans, coffee, tomato soup and body fluids, including blood serum. Of these the greatest proportion were attributed to chloride-induced failures and most of the remainder were from fluoride or caustic contamination.<sup>(26)</sup> In the LMFBR industry all three contaminants may be encountered at some stage of fabrication, shipment, erection and service operations.

### 2.3.1 Effect of Environments on Chloride Stress-Corrosion Cracking

A large amount of work has been conducted on chloride-induced stress-corrosion and a review is given in reference 13. It appears that moisture is an essential requirement for cracking.<sup>(13)</sup>

A standard test for evaluating resistance to chloride-stress corrosion is to expose stressed samples to boiling 42 percent magnesium chloride solution. This is an extremely aggressive environment and it bears little similarity to conditions which might be expected in industry. However, it serves as a useful indication of the relative behavior of alloys which are susceptible to chloride cracking.

Edeleanu<sup>(28)</sup> has shown that in boiling aqueous solutions cracking is accelerated by increasing magnesium chloride

concentration (Figure 17). Other investigations have demonstrated that the presence of oxygen accelerates chloride-induced failure.<sup>(13)</sup>

#### 2.3.2 Effect of Temperature on Chloride Stress-Corrosion Cracking

Several studies have shown that increasing temperature accelerates failures in chloride environments.<sup>(13)</sup> The time to failure ( $t_f$ ) appears to be related to temperature by an expression of the form:

$$t_f = A \exp (B/T)$$

in which  $T$  is the absolute temperature and  $A$  and  $B$  are positive constants.

#### 2.3.3 Effect of pH Value on Chloride Stress-Corrosion Cracking

All of the available data indicate that increasing the pH value leads to increased time to failure in a given chloride environment under a particular set of conditions for applied stress, chloride concentration and temperature.<sup>(13)</sup>

#### 2.3.4 Effect of Alloy Content on Chloride Stress-Corrosion

Work by Copson<sup>(29)</sup> indicates that for increasing concentrations above approximately 10 percent, nickel leads to enhanced resistance of iron-chromium-nickel alloys to stress corrosion (see Figure 18). In the permissible range of nickel concentrations for Type 304 and Type 316 stainless steel (8-10 and 10-14 percent, respectively), the data are not explicit but it does appear that Type 316 stainless steel should resist stress-corrosion more effectively. The data in Figure 19 clearly confirms this. These same data indicate that for regular and low-carbon grades of Types 304 and 316 stainless steel the resistance to cracking is

similar. However, there is some indication that for very high carbon concentrations, above about 0.1 weight percent, increased resistance to stress corrosion is likely (see Figure 20). Nitrogen additions appear to have the reverse effect and enhanced susceptibility to stress corrosion has been reported.<sup>(32, 33)</sup>

#### 2.3.5 Effect of Structure on Chloride Stress Corrosion Cracking

The effects of sensitization on cracking have been investigated but the data are conflicting. Scharfstein and Brindley<sup>(34)</sup> found that for a given exposure in a chloride solution sensitized Type 304 showed deep intergranular cracks. As-received material showed shallower transgranular cracks. Rideout<sup>(35)</sup> confirmed the deleterious effects of sensitization. Other workers, however, find that for the austenitic stainless steels the degree of sensitization does not noticeably affect chloride cracking,<sup>(30, 36, 37)</sup> with the exception of Type 304 L which appeared to be adversely affected by sensitization.<sup>(37)</sup>

#### 2.3.6 Stress Corrosion in Caustic Environments

Compared to chloride-induced cracking relatively little work has been conducted on caustic stress corrosion. This type of failure is important in an LMFBR system where caustic contamination may occur in the intermediate sodium system because of leakage in the steam generator. Recent work<sup>(38)</sup> on Types 304 and 304 L stainless steel has studied the nature of cracking in 10 and 50 percent sodium hydroxide solutions at 600°F. For "U-bend" samples made from material in the as-received and the solution-treated conditions both the 304 and 304 L grades failed in less than 3 days in the 10 percent solution and in less than one day in the 50 percent solution. Under the inspection procedures used, no differences could be detected between the two grades of steel. In the 10 percent solution cracking was intergranular but for the 50 percent solution it was transgranular.

Aging of the as-received and solution-treated materials for periods greater than about 100 hours in the sensitization range were effective in inhibiting cracking in the 10 percent solution but had no apparent effect in the 50 percent solution. Work by Agrawal and Staehle<sup>(39)</sup> indicates, however, that in boiling solutions of sodium hydroxide at ambient pressures sensitization is detrimental in Type 304. Reasons for the discrepancy are as yet unresolved.

## 2.4 Sodium Corrosion Rates

In the high-temperature regions of flowing sodium systems austenitic stainless steels are susceptible to the loss of metallic elements such as chromium, nickel and manganese.<sup>(13)</sup> The loss of nickel and manganese, which stabilize austenite, results in the formation of a ferritic surface layer which will be substantially weaker than the bulk material. As the ferritic layer gets thicker the rate at which the metallic elements diffuse through it becomes slower and a steady state is reached when the proportions of elements removed by the sodium is approximately equal to the chemical composition of the steel. Estimates have been made for the steady-state ferrite layer thickness for an all-austenitic system at a maximum temperature of 1100°F.<sup>(40)</sup> For an oxygen content of < 10 ppm, as measured by the amalgamation technique, and a sodium flow rate of 7 to 20 feet per second, the thickness is about 0.13 mils. This will not cause a measurable effect on the overall strength of an LMFBR component.

Estimates have also been made for the corrosion rates of austenitic stainless steels. For conditions similar to those given above a steady-state rate of 0.08 mils/year was obtained.<sup>(40)</sup> This translates to a net corrosion loss of 2.4 mils over a 30-year lifetime for a component at 1100°F. For the components listed in Section 1 the corrosion losses would be smaller because of their lower operating temperatures.

## 2.5 Fabrication and Weldability

A detailed review of fabricability and weldability of Type 304 and Type 316 stainless steel is given in reference 13. Differences in behavior between the two are relatively minor and a selection of one material over the other based on these properties would be difficult.

Differences do exist, however, between the high, low and normal carbon grades. As stated in Section 2.2, sensitization problems are greatly minimized for low-carbon alloys which allows greater latitude in stress relieving heat treatments which may be required. Welding problems are also decreased for the same reason. The order of preference, for minimizing welding difficulties, is 304 L, 304, 316 L and 316. <sup>(13)</sup>

## 3. PROPERTIES OF FERRITIC 2 1/4 Cr-1Mo STEEL

2-1/4 Cr-1 Mo steel is used at temperatures up to 1200°F in situations where high corrosion resistance is not required. Table 7 gives the chemical composition limits for this material. The chromium content improves corrosion resistance and the molybdenum increases the yield, tensile and creep strengths.

If this alloy is rapidly cooled from hot forming or welding temperatures a hard brittle phase called martensite is formed which greatly decreases ductility. Heat treatment is required to restore some of the ductility. Two basic annealing treatments may be used. <sup>(41)</sup>:

- a) Soak material at 1500 to 1700°F and slow cool.
- b) Soak material at 1200 to 1400°F and air cool.

The length of time required for soaking increases with increasing section thickness.

In order to use the material in a higher strength condition it may be either normalized and tempered or quenched and tempered. For the first case normalization involves annealing in the range 1650 to 1750°F followed by air cooling.

Tempering is achieved by reheating to 1200 to 1400°F followed by air cooling.<sup>(41)</sup>

For the second heat treatment water-dip-quenching from about 1750°F is followed by tempering in the range 1000 to 1200°F followed by air cooling<sup>(42)</sup>

Properties of 2 1/4 Cr-1Mo are given in the following sections:

### 3.1 Mechanical Properties.

The mechanical properties of 2-1/4 Cr-1 Mo are reasonably well documented but, because of the wide range of practical heat treatments which may be used, substantial variations in properties are noticed. Carbon levels are usually close to the permissible maximum of 0.15 weight per cent because of the enhanced strength. A recent publication, however, has investigated the effects of low carbon levels on the mechanical behavior<sup>(43)</sup>. A limited amount of work has also been formed on the effects of in-sodium interstitial transfer on properties.<sup>(5, 6)</sup> Effects of variations in nitrogen content have not been investigated.

#### 3.1.1 Short-Term Tensile Properties

Figure 21 and 22 show typical tensile properties for material in the annealed, normalized and tempered, and quenched and tempered conditions. Note the sharp drop in strength above 1000°F and the ductility minimum at 800°F which are typical of all three heat treatments. Information on thermal aging shows that losses in strength will occur although ductility remains essentially constant<sup>(5, 41)</sup>.

In Figure 23 the effects of carbon on yield and tensile strength are given<sup>(43)</sup>. For a heat of material cast with a carbon content of 0.020 weight percent the strength is similar to that for standard material. However, for material decarburized to 0.021 weight



per cent a large strength loss is noticed. The difference in behavior for the two heats is believed to be connected with grain size.

Information on in-sodium decarburization shows that, for normalized and tempered material, large strength decreases and ductility increases are likely.<sup>(5)</sup>

### 3.1.2 Stress-Rupture Properties

Figure 24 compares the  $10^3$ ,  $10^4$  and  $10^5$  hour rupture strengths for annealed and normalized and tempered 2-1/4 Cr-1 Mo.<sup>(74)</sup> The normalized and tempered material is significantly superior to annealed material. Some information on the 1100°F stress rupture properties of normalized and tempered alloys, as a function of carbon, is summarized in Figure 25. For a given stress, decreasing the carbon level from 0.13 to 0.027 weight percent will decrease the rupture time by an order of magnitude. This is equivalent to a reduction in rupture strength of about 40 percent (see Figure 26).

### 3.1.3 Fatigue Properties

In Table 8 are shown the low-cycle fatigue properties of standard and decarburized materials at 900 and 1100°F for a total strain range of 0.005 in/in at a frequency of 0.25 Hz. For low-carbon concentrations a drastic decrease in fatigue life may be anticipated.

## 3.2 Stress Corrosion Cracking

There do not appear to be any published data which clearly show that unstabilized 2-1/4 Cr-1 Mo steel is susceptible to stress-corrosion cracking in chloride or caustic environments.

## 3.3 Sodium Corrosion Rates

The corrosion rates of 2-1/4 Cr-1 Mo have been measured by several groups over a wide range of temperatures and oxygen levels. Figure 27 summarizes the available information. None of the oxygen analyses



were apparently measured by the currently favored vanadium wire technique. However, with the exception of the 0.5 ppm oxygen analysis, which was obtained by the plugging meter technique, the 1100°F corrosion rates for 2 1/4 Cr-1 Mo appear to be higher than those for Types 304 and 316 stainless steel (see Section 2.4). The work of Sannier et al<sup>(75)</sup> directly compared the corrosion rates of 2 1/4 Cr-1 Mo and Type 304 stainless steel for exposure periods of about 6000 hours at 1200°F. The data, given in Figure 28, show that 2 1/4 Cr-1 Mo has a much faster initial corrosion rate although at longer times the steady state rates for the two groups of alloys may not be significantly different. Hence, for LMFBR coolant-boundary components the metal corrosion loss at maximum operating temperatures is not likely to be a problem over the 30-year lifetime of the components.

### 3.4 Steam-Corrosion Rates

The data presented in Figures 29 and 30 show the steam corrosion rates for 2 1/4 Cr-1 Mo for temperatures between 950 and 1200°F. Specimen conditions, steam pressures and boiler water chemistries are given in Table 9. All exposures were under isothermal conditions. The gradient of the 950°F line was predicted by assuming that an Arrhenius type oxidation rate was applicable, thus enabling an extrapolation of the oxidation rates for the 1100 and 1000°F tests. Conclusions obtained from these data include:

- a) Within normal experimental scatter there is no difference between the corrosion rate of as-machined, pickled, and pickled and stress-relieved surfaces.
- b) Within the range of steam pressures studied the corrosion rates appear to be generally independent of pressure.
- c) There is a large difference in the corrosion rates between Oak Ridge National Laboratory and ASME samples at both 1100 and 1200°F. One possible explanation is that the ORNL tests

involved reducing the specimen temperatures to room temperature every 1000 hours in order to expose them intermittently to saturated steam. This could possibly cause cracking of the protective oxide scale which would lead to accelerated corrosion.

d) At 1200°F the rate of corrosion is markedly accelerated.

Table 10 gives estimates of the metal loss for 30-year exposures of 2-1/4 Cr-1 Mo to steam.

### 3.5 Sodium-Water Reaction Effects

An early study compared the rates of wastage for 2-1/4 Cr-1 Mo, Incoloy 800 and Type 316 stainless steel during exposure to a sodium/water reaction.<sup>(51)</sup> This type of information is most important in selecting LMFBR steam generator tubing material because of the possibility of a water-to-sodium or steam-to-sodium leak. Table 11 gives the wastage rates for these tests. The testing technique involved the injection of water into a pot containing a 3 lb. mass of sodium at 930°F. The water was injected at right-angles to a plate sample which was weighed before and after a 30 minute test. The 2-1/4 Cr-1 Mo was greatly inferior in performance to the austenitic alloys. More recent studies have confirmed the poor resistance of 2-1/4 Cr-1 Mo to caustic corrosion.<sup>(52)</sup>

### 3.6 Fabrication and Weldability

#### 3.6.1 Hot Forming

Suggested temperatures for bending are 1750 to 1850°F and for forging, 2000 to 2200°F. Stress relieving should be conducted at 1300 to 1400°F and annealing should be at about 1000°F followed by slow cooling.

Some ferritic steels, including 2-1/4 Cr-1 Mo, may, under certain conditions, be susceptible to temper embrittlement and stress-relief cracking. These phenomenon are, as yet, incompletely

understood but procedures have been developed to minimize their occurrence during fabrication and subsequent component service. A detailed review of these and associated embrittlement problems for 2 1/4 Cr-1 Mo is given in reference 53.

### 3.6.2 Welding<sup>(54)</sup>

2 1/4 Cr-1 Mo is readily weldable by the shielded metal-arc, submerged-arc, gas metal-arc, flux-cored-arc, gas tungsten-arc and electroslag processes. Since this alloy is air-hardenable, care must be exercised to prevent cracking of the weld and heat-affected zone materials. Minimum pre-heat temperatures of 300°F are usually specified to reduce the possibility of cracking by:

- a) Allowing hydrogen to escape, if it is present.
- b) Reducing the rate of cooling from the welding temperature.

Filler metals containing 2 1/4 Cr-1 Mo may be used although United States Steel Corporation have recommended 5 Cr-1/2 Mo.<sup>(41)</sup>

Post-weld heating to 1100 to 1375°F is usually recommended for stress relief.

### 3.6.3 Welding of 2 1/4 Cr-1 Mo to Austenitic Stainless Steels

The joining of ferritic to austenitic materials poses special problems because of the different thermal expansion rates; see Table 12. In addition, carbon will tend to diffuse from the higher-carbon ferritic material during service into the weld metal which will have a composition dependent on the proportions

of melted ferritic, austenitic and filler metals. Hence, problems may arise from the higher carbon content which will occur in the austenitic material, or the depleted carbon level in the ferritic steel, after long-term service at high temperature. Such weld failures have continued to recur even after 20 years' of experience.<sup>(57-59)</sup> Filler metals which have been satisfactorily used are: Type 307, 16-18-2, Inco 182<sup>(60-63)</sup> (i.e. ENiCrFe-3). Chemical compositions of these alloys are given in Table 13. To ensure dimensional stability, dissimilar metal joints should be heated to 100°F above the maximum operating temperature and slow cooled.<sup>(13)</sup>

A detailed welding procedure for welding Type 304 stainless steel pipe to 2 1/4 Cr-1 Mo has recently been described.<sup>(64)</sup> Because of its direct relevance to the current project, the description is reproduced below in its entirety:

#### Nickel Alloy Filler Metal For Crack-Free Welding

The selection of a filler metal was an important factor in welding a 2 1/4 Cr-1 Mo steel pipe (ASTM A387, grade D) to a Type 304 stainless steel pipe. This pipe assembly was a component of a steam pipeline that had to withstand cyclic heating and cooling between 1050°F and room temperature. Details of the joint between the 15-in.-OD by 1-in. wall pipes before and after welding, in 22 passes, by the shielded metal-arc process are shown in Figure 31. A nickel alloy flux-covered electrode (ENiCrFe-3) was selected in preference to an austenitic stainless steel electrode, for three reasons:

- a) The coefficient of thermal expansion (9.5 micro-in. per inch per °F) of weld metal from the ENiCr Fe-3 electrode is close to that of the ferritic 2 1/4 Cr-1 Mo steel. Thus, during cyclic temperature service, the major differential expansion stresses developed primarily at the stronger interface between the stainless steel pipe and the weld metal, rather than at the weaker interface between the ferritic steel pipe and the weld metal - the location at which such expansion stresses would have developed if a stainless steel electrode had been used.
- b) Carbon depletion in the ferritic steel was less when the nickel alloy weld metal was used than when a stainless steel weld metal was used. Therefore, the heat-affected zone of the ferritic steel was not weakened by loss of carbon.
- c) The excellent deposition characteristics of the nickel alloy electrode produced a sound, porosity-free weld that was not subject to weld-metal cracking.

The pipe ends were machined to the joint configuration shown in Figure 31 and wiped clean with a solvent; then the pipes

were mounted horizontally on turning rolls for welding. Low-frequency induction heating was used to provide a 500°F preheat for the ferritic pipe. This temperature was maintained, by automatic control, throughout the welding operation.

The joint was tack welded at 6-in. intervals, using a 1/8-in. diameter electrode and the same current and voltage settings as for the first pass (see table of welding conditions with Figure 31. The edges of the tack welds were thinned by grinding before the root pass, which was carefully deposited to obtain a smooth inside root surface. Each weld bead was deslagged and visually inspected before the next bead was deposited. Immediately after welding, the weld was postheated, by induction heating, to 1350°F for 1 hour and air cooled.

Welds made with the ENiCr Fe-3 electrode met bend-test and tension-test requirements of Section IX of the ASME Boiler and Pressure Vessel Code, and also passed visual, dye-penetrant, radiographic and metallographic inspection. Fractures produced in transverse tension tests at room and elevated temperatures were in the base metal away from the heat-affected zones. The welds had markedly better resistance to cracking during thermal cycling than did similar joints welded with Type 347 stainless steel electrodes.



#### 4. INTERSTITIAL TRANSFER EFFECTS

The transfer of interstitials, such as carbon and nitrogen, in LMFBR components is a potential problem because of the attendant changes in mechanical strength during long-term service (see Sections 2.1 and 3.1, above). Conceivably, a component meeting minimum strength and interstitial content requirements at start-of-life could violate these requirements at some point during service.

To date, there are no proven techniques which will accurately predict the rates of interstitial transfer and end-of-life mechanical behavior under complex LMFBR conditions. However, by using various simplifying assumptions attempts have been made at ARD to derive estimates of the extent of interstitial transfer.<sup>(66-68)</sup>

##### 4.1 Selection of Carbon Potential Values

In order to estimate the rates of interstitial transfer in the CRBRP primary and intermediate systems, values for the carbon potential ( $C_s$ ) need to be established. ARD is currently assuming that the carbon potential for both systems is 30 ppm. Reasons for this selection are given below:

###### a) Primary Side

ARD had carried out a survey of the carbon potentials of operating test loops fabricated of unstabilized 300 series stainless steels, including EBR-II.<sup>(69)</sup> Generally, the survey indicated that the carbon potentials were low and of the order of 30 ppm. The 30 ppm  $C_s$  value has been used for FFTF predictions<sup>(66)</sup> and is likewise considered to be the most representative value for estimation purposes here.

###### b) Intermediate Side

Early work at G.E.<sup>(70)</sup> and some observations made by B & W<sup>(71)</sup> indicate that when a system contains a ferritic alloy, such



as 2-1/4 Cr-1 Mo, the carbon lost from the ferritic alloy will cause carburization of the stainless steel in the system. Recent G.E. work, however, would suggest that decarburization rates of 2-1/4 Cr-1 Mo are much lower than previously expected<sup>(72)</sup>. These observations are supported by measurements made at ARD on the diffusion of carbon from 2-1/4 Cr-1 Mo<sup>(69)</sup> and by French workers who saw virtually no decarburization below 900°F after 6000 hours<sup>(43)</sup>. However, no previous estimates of  $C_s$  have been made for bimetallic (austenitic/ferritic) systems. Until direct measurements are made, therefore, it will be assumed that for the intermediate system,  $C_s$  is equal to 30 ppm.

#### 4.2 Compensation for Nitrogen Transfer

It is known that nitrogen has at least as great an effect on the mechanical properties of austenitic steels as carbon<sup>(1)</sup>. It is also known that nitrogen will transfer in a manner similar to that of carbon. The following assumptions are made to evaluate the additional effects of nitrogen transfer:

- a) The nitrogen content is assumed to be equal to that for carbon.
- b) The  $C_s$  versus  $C_e$  relationships for carbon ( $C_e$  is the carbon content of the stainless steel at the sodium interface) are assumed to be applicable to nitrogen.

These assumptions imply that in all subsequent calculations the effective interstitial content ( $C + N$ ) is taken to be  $2C$ , and that nitrogen is assumed to behave indentically to carbon.

#### 4.3 Calculation of Interstitial Gradients

In order to obtain interstitial concentration gradients in components exposed to sodium for a design lifetime of 250,000 hours (~ 30 years) a computer program was used to solve Fick's second law of diffusion<sup>(66)</sup>.

This states:

$$\frac{\partial c}{\partial t} = D \frac{\partial^2 c}{\partial x^2}$$

in which  $c$  = interstitial concentration, i.e.  $(C + N)$   
 $t$  = diffusion time  
 $x$  = depth below sodium/metal interface  
 $D$  = effective interstitial diffusion coefficient

This equation was solved using the boundary conditions

$$x = 0 \text{ at } c = C_e$$

and  $x \rightarrow a \text{ at } \frac{\partial c}{\partial x} = 0$

Values of  $C_e$  and  $D$  were obtained from reference 66. Component temperatures used in the calculation are given in Table 14 together with materials which were studied in the computer analysis. Typical values of interstitial content have been assumed, as shown. Results of the study are given below.

#### 4.3.1 Primary System Hot-Leg Components

Hot-leg components at 995°F undergo severe decarburization and denitridation at the surfaces (Figure 32). At the sodium/metal interface the total interstitial concentration is about 160 ppm for Types 316 and 316 H and about 230 ppm for Types 304 and 304 H. These are below the minimum permissible carbon concentration of 0.04 weight percent required for high-temperature design<sup>(3)</sup>.

Significant differences exist between Type 304 and Type 316. Because of the slower interstitial diffusion rates for Type 316<sup>(3)</sup> the depth of interstitial depletion is only about 60 percent of

that for Type 304. The end-of-life depths of interstitial depletion, as a percentage of component thickness, are given in Table 14 in order to illustrate the severity of interstitial transfer for components of varying thicknesses. The results discussed in this section also apply to the auxiliary system hot leg because of the similar operating conditions.

#### 4.3.2 Primary System Cold-Leg Components

The data in Figure 33 show that for Types 304 and 304 L a heavily carburized and nitrided surface will result from a 30-year exposure to sodium. The surface (C + N) concentration for these alloys is about 1460 ppm. Since the carbon concentrations are one-half of these values it can be seen that the surface material of Type 304 L exceeds the permissible carbon concentration maximum of 0.03 weight percent<sup>(3)</sup>. Normal Type 304, however, falls within the 0.08 weight percent limit<sup>(3)</sup>. Carburization depths are given in Table 14.

#### 4.3.3 Intermediate System Hot-Leg Components

The data for the hot-leg components in the intermediate system (Figure 34) are analogous to those for the primary system. Types 316 and 316 L steel are superior to the equivalent Type 304 grades in resisting interstitial loss. Compared with the primary system, the depth of interstitial depletion is much smaller.

#### 4.3.4 Intermediate System Cold-Leg Austenitic Components

The data given in Figure 35 and Table 14 for austenitic components are similar to those for the primary system. Again heavy surface carburization and nitridation are likely although the depth of diffusion is small. For both the primary and intermediate system cold-leg components, the heavily carburized and nitrided surface layer will be far less ductile and under fatigue or bending loads, surface cracking may be induced.<sup>(12)</sup>

#### 4.3.5 Intermediate System Ferritic Components

The rates of interstitial transfer for 2-1/4 Cr-1 Mo cannot currently be predicted using the same type of computer study as for the austenitic stainless steels. From Section 4.1 it seems that any transfer will not significantly affect steam generator components except possible, the tubing. Conclusions reached from ARD studies are<sup>(69)</sup>:

- a) The rate of carbon loss from 2-1/4 Cr-1 Mo will proceed at much the same rate as for Type 304 stainless steel.
- b) At temperatures below 1100°F, carbon depletion is unlikely to fall below an equilibrium level of 500 ppm regardless of the prevailing carbon activity in the sodium.
- c) At temperatures below 900°F, data extrapolation indicates that the material will have a tendency to carburize even with relatively low carbon activity levels in the sodium. Carburization, however, will not exceed 2000 ppm.

More recent work, however, indicates that Conclusion (c) may not be valid. It is now thought likely that below 900°F carburization may not, in fact, occur.<sup>(76)</sup>

#### 4.4 Comparison Between Low, Normal and High-Carbon Grades

For all austenitic components situated in the hottest regions of the primary and intermediate system hot legs, significant losses of interstitials will occur from the sodium-exposed surfaces. However, the bulk carbon concentrations will remain above the 0.04 weight percent minimum<sup>(3)</sup> if the starting (C + N) levels are similar to those specified in this study (Table 14). The Type 316 grades resist decarburization and denitridation much more effectively than Type 304. Hence, if maximum mechanical strength is required, Type 316 would be the optimum choice. The presence of the low-interstitial surface layer should present no problems from a mechanical standpoint because of its enhanced ductility.

In cold-leg regions Type 316 also resists carburization and nitridation more effectively than Type 304. At these lower temperatures, however, mechanical strength is a less serious design problem. Thus, the selection of either Type 304 or Type 316 would be appropriate. For maximum resistance to carburization and nitridation, Type 316 would be the more suitable of the two. Note that for the lower temperature regions of the cold-leg, where large interstitial gradients are present, it is possible to minimize the gradients by selecting a higher carbon grade of steel. In fact, since the  $C_e$  value is dependent only on temperature for a given  $C_s$ , there will be zero interstitial gradient at a location in the system where  $C_e$  is equal to the alloy interstitial level ( $C + N$ ).

## 5. COST CONSIDERATIONS

A detailed evaluation of the relative costs of fabricating Type 304 versus Type 316 components cannot currently be made. Although estimates were obtained on the costs of starting materials, this would only be a fraction of the component cost because of additional fabrication, welding, inspection and testing. An estimate of these costs is beyond the scope of this study.

To assess the differential between Types 304 and 316 starting alloys, estimates for plate material were considered<sup>(73)</sup>. However, the differential was essentially the same for all product forms. Table 15 gives the costs of Types 304, 304 L, 316 and 316 L alloys. For quantities over 10,000 pounds the cost per pound is decreased slightly. No prices were available for high-carbon grades of steel since they would have to be made to order.

Material purchased to RDT Standards will increase the base price by 25 percent whereas material meeting ASME requirements for Section III Class 1 components will increase the base price by 10 percent. Any ultrasonic testing requirements will add an additional increment of 10 percent on the base price.

Compared to austenitic stainless steels, 2-1/4 Cr-1 Mo starting material is less expensive. However, the greater degree of difficulty in welding, fabricability and heat treatment may make a ferritic component almost as costly as a stainless steel one.

## 6. GUIDELINES FOR ALLOY SELECTION

The following guidelines may be used in selecting a particular alloy for a specific LMFBR application. Table 16 summarizes these guidelines.

### 6.1 Selection for Interstitial Transfer and Mechanical Behavior

In Section 4, above, it was shown that in hot-leg components made from regular and high-carbon materials the carbon concentrations at the sodium-exposed surfaces could be reduced below ASME Code-specified minimums during service, although bulk carbon concentrations in heavy-section materials would not violate these minimum values<sup>(2, 3)</sup>. Note that the equilibrium interstitial concentration at the sodium-exposed surface of a component is dependent only on temperature for a given  $C_s$  value (see Section 4.4) so that it is not possible to avoid low surface carbon levels by choosing a high-carbon steel. In addition, the effects of nitrogen, which exert as large an effect as carbon on mechanical strength<sup>(1)</sup>, are not factored into the Code. It is recommended that in ordering any austenitic stainless steel for use in the CRBRP, the nitrogen levels be at least as high as those for carbon and that interstitial contents be in the higher range of permissible concentrations.

The Type 316 and 316 H grades of steel are superior to Type 304 and 304 H, especially in rupture strength and tensile ductility. If maximum strength is an overriding design criterion, Type 316 H would be the optimum material to choose since it would limit the minimum permissible carbon level to 0.04 weight percent (see Table 1).

In the cold-leg, components could probably be made from normal grades of either Type 316 or Type 304 since at these lower temperatures mechanical strength is a less stringent design requirement. To differentiate between Type 304 and Type 316 a trade-off study would be required to compare the lower cost, better weldability and fab-



ricability, and increased resistance to sensitization of Type 304 with the superior strength and resistance to stress-corrosion of Type 316.

Below about 850°F the  $S_{mt}$  allowable design stresses for 2-1/4 Cr-1 Mo are larger than those for the Types 304 and 316 stainless steel. At lower temperatures, therefore, if there is a cost incentive, this alloy may possible be used instead of stainless steel.

## 6.2 Selection for Fabricability, Weldability and Sensitization

For components in lower-temperature regions where the highest-strength grades of steel are not mandatory, significant benefits will result in the selection of regular, rather than high-carbon, grades of Type 304 or Type 316. Welding problems are minimized, and any required stress relieving operations are simplified since sensitization is significantly delayed in lower carbon materials. The weldability of 2-1/4 Cr-1 Mo is inferior to that of austenitic stainless steel in view of the requirement for pre- and post-weld heating. However, this material does not suffer from sensitization problems which are common for the stainless steels.

Regarding the potential difficulties of dissimilar metal welds in the Demonstration Plant intermediate system, it would be advisable to minimize the number to be used in order to limit potential problems due to different thermal expansion rates for austenitic and ferritic materials, local interstitial transfer, the possibility of sensitizing the austenitic material during post-weld heat treatment of the weld, and weld repairability.

## 6.3 Selection for Stress-Corrosion Cracking Resistance

Type 316 stainless steel appears to be more resistant to chloride-induced cracking than Type 304, based on the data in Figure 19. However, the problem is complicated by so many variables that it may not be superior under all LMFBR conditions. Hence, the apparent inferiority of Type 304 should not be used as an argument for disqualifying its use in a component.

It is not possible, at this time, to identify the stainless steel for best resistance to caustic-induced stress corrosion.



Compared to the austenitic stainless steels, 2-1/4 Cr-1 Mo is substantially more resistant to halide and caustic stress corrosion.

#### 6.4 Tentative Recommendations for Material Selection

Table 17 gives tentative recommendations for the selection of materials for CRBRP components. These recommendations are based primarily on metallurgical considerations. If design calculations indicate that the highest-strength materials are required, then H grades of steel may be selected in order to ensure that purchased material does not contain less than 0.04 weight percent carbon. Nevertheless, whether the normal or H grade of steel is ordered it is recommended that the acceptable range of carbon concentration be specified in the ordering data.

The use of low-carbon grades of stainless steel are restricted to temperatures  $\leq 800^{\circ}\text{F}^{(2)}$  and, therefore, cannot be used in hot-leg components in the coolant boundary. Their lower strength also limits their usefulness in cold-leg components.

In certain situations, where there is a cost-incentive, 2-1/4 Cr-1 Mo may be used in lower temperature coolant-boundary areas in place of stainless steel.

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TABLE 1  
Carbon Contents for Austenitic Stainless Steel Product Forms\*

ASME MATERIAL SPECIFICATION	CARBON CONTENT (ppm)		
	304 L and 316 L SS	304 and 316 SS	304 H and 316 H SS
SA 240; Plate, Sheet and Strip For Fusion Welded Unfixed Pressure Vessels	$\leq 0.03$	$\leq 0.08$	-
SA 249; Welded Austenitic Stainless Boiler Superheater and Heat Exchanger and Condenser Tubes	$\leq 0.03$	$\leq 0.08$	-
SA 312; Seamless and Welded Austenitic Stainless Steel Pipe	$\leq 0.035$ $\leq 0.040$ for small diameter pipe	$\leq 0.08$	0.04 - 0.10
SA 376; Seamless Austenitic Stainless Steel Pipe For High Temperature Central Station Service	-	$\leq 0.08$	0.04 - 0.10
SA 479; Heat Resisting Steel Bars and Shapes for Boilers and Other Pressure Vessels	$\leq 0.03$	$\leq 0.08$	-

\* Data from Reference 2

TABLE 2

S<sub>y</sub> and S<sub>m</sub> Values for Austenitic Stainless Steels\*

TEMPERATURE (°F)	S <sub>y</sub> (ksi)				S <sub>m</sub> (ksi)			
	304 L	304 and 304 H	316 L	316 and 316 H	304 L	304 and 304 H	316 L	316 and 316 H
100	25.0	30.0	25.0	30.0	16.6	20.0	16.6	20.0
200	21.3	25.0	21.1	25.8	16.6	20.0	16.6	20.0
300	19.1	22.5	18.9	23.3	16.6	20.0	16.6	20.0
400	17.5	20.7	17.2	21.4	15.7	18.7	15.5	19.2
500	16.3	19.4	15.9	19.9	14.7	17.4	14.4	17.9
600	15.5	18.2	15.0	18.8	13.9	16.4	13.5	17.0
700	14.9	17.7	14.3	18.1	13.4	15.9	12.8	16.3
800	14.4	16.8	13.7	17.6	13.0	15.1	12.3	15.8
900	13.9	16.2	13.1	17.3		14.6**		15.7**
1000	13.3	15.6	12.4	17.0		14.0**		15.5**
1100						13.3**		14.8**
1200						12.7**		14.6**

\* Data are from ASME Boiler and Pressure Vessel Code, Section III and ASME Code Case 1331-8, and apply to Class 1 components

\*\* For Type 304 and Type 316 grades only

TABLE 3  
Corrosives Reported to Induce Intergranular Corrosion  
in Sensitized Austenitic Stainless Steel<sup>(65)</sup>

Acetic acid	Oxalic acid
Acetic acid + salicylic acid	Phenol + naphthenic acid
Ammonium nitrate	Phosphoric acid
Ammonium sulfate	Phthalic acid
Ammonium sulfate + H <sub>2</sub> SO <sub>4</sub>	Salt spray
Beet juice	Sea water
Calcium nitrate	Silver nitrate + acetic acid
Chromic acid	Sodium bisulfate
Chromium chloride	Sodium hydroxide + sodium sulfide
Copper sulfate	Sodium hypochlorite
Crude oil	Sulfite cooling liquor
Fatty acids	Sulfite solution
Ferric chloride	Sulfite digester acid
Ferric sulfate	(calcium bisulfite + sulfur dioxide)
Formic acid	Sulfamic acid
Hydrocyanic acid	Sulfur dioxide (wet)
Hydrocyanic acid + sulfur dioxide	Sulfuric acid
Hydrofluoric acid + ferric sulfate	Sulfuric acid + acetic acid
Lactic acid	Sulfuric acid + copper sulfate
Lactic acid + nitric acid	Sulfuric acid + ferrous sulfate
Maleic acid	Sulfuric acid + methanol
Nitric acid	Sulfuric acid + nitric acid
Nitric acid + hydrochloric acid	Sulfurous acid
Nitric acid + hydrofluoric acid	Water + starch + sulfur dioxide
	Water + aluminum sulfate

TABLE 4 (a)

Effect of Annealing Temperature on Reduction of Stress in  
Austenitic Stainless Steel<sup>(17)</sup>

Annealing Temperature	Effect on Stress
400 - 750°F	about 5% general relief, but reduces peak stresses by up to 40%
1000 1200°F	about 35% relief
1550 1650°F	about 85% relief
1750 1900°F	maximum relief of about 95%

TABLE 4 (b)

Effect of Annealing Temperature and Time on Reduction of Stress in Type 347 Pipe<sup>(18)</sup>

Annealing Temperature	Residual Stress
(Size 5 in. O.D., 4 in. I.D.)	
As Welded	15,000 to 18,500 psi on I.D.
1200°F (4 hours)	13,700 to 15,300
1200°F (12 hours)	16,000
1200°F (36 hours)	15,600
1650°F (2 hours)	0
1850°F (1 hour)	0
(Size 9-1/4 in. O.D., 6-1/2 in. I.D.)	
As Welded	26,000 to 30,000 psi on I.D.
1100°F (16 hours)	20,000
1100°F (48 hours)	20,000
1100°F (72 hours)	23,000
1200°F (4 hours)	21,500 to 24,000

TABLE 4 (c)

Effect of Annealing Temperature on Relief of Stress in Type  
316 Cold Drawn Tubing <sup>(19)</sup>

<u>Annealing Temperature</u>	<u>Measured Stress, Circumferential</u>
No heat treatment, 1/4 hard	52,500 psi
1000°F (24 hours, air cool)	44,500
1200°F (1/2 hour, air cool)	39,100
1375°F (1/2 hour, air cool)	26,700
1375°F (1/2 hour, furnace cool)	23,100
1450°F (1/2 hour, furnace cool)	10,400
1550°F (1/2 hour, furnace cool)	3,600
1600°F (1/2 hour, furnace cool)	3,500



TABLE 5  
Results of Huey Tests on Several Heats of Nitrogen-Alloyed Stainless Steel (21)

C %	N %	PLATE THICKNESS (IN)	CORROSION RATE	
			ANNEALED (mils/yr)	SENSITIZED (mils/yr)
0.035	0.070	9-1/2	11.5	--
			10.6	--
			8.9	--
			8.5	--
			7.0	--
.066	0.140	7-3/4	10.0	--
			12.2	--
			12.2	--
.056	0.089	7	7.8	--
.063	0.110	6-1/16	8.5	--
			7.2	--
.057	0.110	5-3/4	10.3	35.6
			10.0	31.4
			10.3	33.1
			11.6	--
.030	.100	3	6.0	13.2
.025	.130	3	6.0	7.2
.070	.040	3	6.0	84.0

TABLE 6

Huey Test Data for AISI 304 and Jones &amp; Laughlin

304-N (0.25 Nitrogen)<sup>(22)</sup>

		Corrosion Rate in Mils/yr 48 hour periods				
Material	Grain Size	1	2	3	4	5
<u>Cold Rolled Mill Annealed</u>						
Type 304-N	11	8	7	7	7	7
	11	<u>8</u>	<u>8</u>	<u>8</u>	<u>8</u>	<u>9</u>
	Avg.	8	8	8	8	8
Type 304	8	14	12	13	13	13
	8	<u>12</u>	<u>11</u>	<u>11</u>	<u>12</u>	<u>12</u>
	Avg.	13	12	12	13	13
<u>Sensitized for One Hour at 1250°F</u>						
Type 304-N	11	66	128	173	196	
	11	72	136	179	202	
	11	<u>72</u>	<u>139</u>	<u>180</u>	<u>201</u>	
	Avg.	70	134	177	200	
Type 304	8	58	281	454	492	
	8	<u>61</u>	<u>313</u>	<u>479</u>	<u>513</u>	
	Avg.	60	297	467	503	
<u>Annealed for 15 Minutes at 2000°F Sensitized for One Hour at 1250°F</u>						
Type 304-N	4	56	44	43	45	
	4	41	42	50	54	
	4	<u>43</u>	<u>50</u>	<u>69</u>	<u>78</u>	
	Avg.	50	45	54	58	
Type 304	2	58	39	34	34	
	2	45	45	35	41	
	2	<u>44</u>	<u>35</u>	<u>40</u>	<u>42</u>	
	Avg.	49	40	37	39	

TABLE 7  
Chemical Composition of Ferritic 2-1/4 Cr-1 Mo Steel (41)

C	-	0.15 weight percent maximum
Mn	-	0.30/0.60
Si	-	0.50 maximum
Cr	-	1.90/2.69
Mo	-	0.87/1.13
P	-	0.03 maximum
S	-	0.03 maximum

TABLE 8

Low Cycle Fatigue Data for 2-1/4 Cr-1 Mo Steel<sup>(43)</sup>

MATERIAL	TEST TEMPERATURE, DEGREE F	SPECIMEN NUMBER	CYCLES TO FAILURE <sup>a</sup>	INDICATION OF CRACKING <sup>b</sup>
Standard (0.12 Wt.% C)	900	1	3484	3460
		3	2736	
		4	2960	3090
		5	3226	3216
			Avg 3102	
	1000	6	3150	3136
		7	4648	4568
		8	4295	
		9	3115	3078
			Avg 3802	
Decarburized (0.02 Wt.% C)	900	D-2-4	1680	1650
		D-2-5	2160	2175
		D-2-6	1869	1905
		D-2-7	2137	
			Avg 1964	
	1000	D-3-1	1961	
		D-3-4	1340	1362
		D-3-5	2221	2211
		D-3-6	2061	2073
			Avg 1897	

<sup>a</sup>Failure taken as 10 percent decrease in maximum load range

<sup>b</sup>By examination of hysteresis loops

TABLE 9  
Test Conditions for Steam Corrosion Tests on 2-1/4 Cr-1 Mo

Symbol	Material Condition	Steam Pressure (psi)	Boiler Water Chemistry	Reference
▲●	Annealed at 1650°F	2000	N/A	46
►	Annealed at 1650°F, as machined surface	2000	N/A	46
■	Annealed at 1650°F, surface machined and then pickled	2000	N/A	46
◆	Annealed at 1650°F, surface machined, pickled and then stress relieved at 1350°F			
●	Probably annealed	2000	N/A	48
▼	Annealed	900	SiO <sub>2</sub> 0.12-0.20 ppm Na <sub>2</sub> SO <sub>4</sub> 5-25 ppm Na <sub>2</sub> SO <sub>3</sub> 0.0-1.0 ppm PO <sub>4</sub> 1.0-5.0 ppm  NaCl 3-10 ppm pH = 10.1-10.7 pp	47
×	Annealed	2000	N/A	49
●	Annealed	2350	N/A	50
●○	Annealed	2000	N/A	48
▷	Annealed at 1650°F, as-machined	10	O <sub>2</sub> <10 ppb SiO <sub>2</sub> 2-5 ppb Cl <50 ppb Fe <10 ppb Resistivity >1 MΩ pH = 9.5	46
□	Annealed at 1650°F, surface machined and pickled	10	"	46
◇	Annealed at 1650°F, surface machined, pickled and stress relieved at 1350°F	10	"	46
▷	Annealed at 1650°F, as-machined surface	650	"	46
■	Annealed at 1650°F, surface machined and pickled	650	"	46
◆	Annealed at 1650°F, surface machined, pickled and stress relieved at 1350°F	650	"	46

N/A = Not available

TABLE 10

Estimated Steam Corrosion Allowances for Ferritic 2-1/4 Cr-1 Mo Steel

Exposure Temperature (°F)	30-Year Corrosion Allowance (mils)
900	2.9
950	4.9
1000	9.3
1050	16.3
1100	30.1
1200	199.8



TABLE 11  
Wastage Rates in Sodium/Water Reactions<sup>(51)</sup>

MATERIAL	TOTAL WT. LOSS (grams)
Type 316	0.015
2-1/4 Cr-1 Mo	0.436
Incoloy 800	0.004
30 Milliliters of water injection over 30 minutes.	
Sodium Temperature 500°C (930°F)      3 Pounds Sodium Charge	

TABLE 12

Coefficients of Linear Thermal Expansion for Type 304 Stainless Steel  
and 2-1/4 Cr-1 Mo Steel

ALLOY	COEFFICIENTS OF THERMAL EXPANSION, $10^{-6}/^{\circ}\text{F}$					
	RT	200	400	600	800	1000
Type 304 SS <sup>(55)</sup>	8.8	9.1	9.4	9.6	9.8	10.2
2-1/4 Cr-1 Mo <sup>(41, 56)</sup>	6.4	6.6	7.0	7.2	7.5	7.8

TABLE 13

Chemical Composition of Filler Metals for 2-1/4 Cr-1 Mo Welds

ELEMENT	COMPOSITION (WT. PERCENT)		
	INCONEL 182	T 307 SS <sup>(b)</sup>	16-8-2 <sup>(c)</sup>
Ni	Balance	9.0 - 10.5	8.0
Cr	13.0 - 17.0	19.5 - 21.5	16.0
C	≤ 0.10	0.07 - 0.15	
Mn	5.0 - 9.5	3.75 - 4.75	
Fe	6.0 - 10.0	Balance	Balance
S	≤ 0.015	≤ 0.030	
P		≤ 0.030	
Si	≤ 1.0	0.25 - 0.60	
Cu	≤ 0.50	-	
Ti	≤ 1.0	-	
Nb	1.0 - 2.5 <sup>(a)</sup>	-	
Co	≤ 0.12	-	
Mo		≤ 0.25	2.0
N		≤ 0.07	

(a) Plus Ta. Ta 0.25 maximum when specified

(b) Westinghouse proprietary material

(c) Approximate composition for main elements (Babcock and Wilcox proprietary material)

TABLE 14 (a)

Diffusion Depths for Hot-Leg Components

COMPONENT	MIN. THICKNESS (INS)	MAX. TEMP. (°F)	INTERSTITIAL DIFFUSION DISTANCE <sup>(a)</sup> (percentage of component thickness)			
			304 <sup>(b)</sup>	304 H	316	316 H
<u>Primary System Hot Leg</u>						
Vessel	2.0	995	2.5	2.6	1.7	1.7
Vessel Outlet Nozzle	0.5	995	9.8	10.2	6.7	6.7
Piping	0.5	995	9.8	10.2	6.7	6.7
Pump Suction Nozzle	0.5	995	9.8	10.2	6.7	6.7
Pump Discharge Nozzle	0.5	995	9.8	10.2	6.7	6.7
Pump Casing	2.0	995	2.5	2.6	1.7	1.7
IHX Inlet Nozzle	0.5	995	9.8	10.2	6.7	6.7
IHX Shell	1.625	995	3.0	3.1	2.1	2.1
Isolation Valve Inlet Nozzle	0.5	995	9.8	10.2	6.7	6.7
Isolation Valve Outlet Nozzle	0.5	995	9.8	10.2	6.7	6.7
Isolation Valve Body	1.5	995	3.3	3.4	2.2	2.2
<u>Intermediate System Hot Leg</u>						
IHX Outlet Nozzle	0.375	936	7.2	7.7	4.5	4.5
Piping	0.5	936	5.4	5.8	3.4	3.4
Steam Generator Inlet Nozzle	0.375	936	-	-	-	-
Steam Generator Shell	0.5	936	-	-	-	-
Isolation Valve Inlet Nozzle	0.375	936	7.2	7.7	4.5	4.5
Isolation Valve Outlet Nozzle	0.375	936	7.2	7.7	4.5	4.5
Isolation Valve Body	0.75	936	3.6	3.9	2.3	2.3
<u>Auxiliary System</u>						
Hot-Leg Piping	0.322	995	15.2	15.9	10.3	10.3
Sample Loop Piping	0.133	995	36.8	38.4	24.8	24.8
Valves	0.216	995	22.7	23.6	15.3	15.3

- (a) The interstitial diffusion distance is arbitrarily taken to be the depth at which the (C + N) concentration has changed from  $C_e$  to within 90 percent of the starting (C + N) concentration.
- (b) Type 304 and Type 316 are assumed to have 0.06 weight percent carbon and 0.06 weight percent nitrogen. Type 304 H and Type 316 H are assumed to have 0.09 weight percent carbon and 0.09 weight percent nitrogen.

TABLE 14 (b)

Diffusion Depths for Cold-Leg Components

COMPONENT	MIN. THICKNESS (INS)	MIN. TEMP. (°F)	INTERSTITIAL DIFFUSION DISTANCE <sup>(a)</sup> (percentage of component thickness)	
			304 <sup>(b)</sup>	304 L
<u>Primary System Cold Leg</u>				
Vessel	2.0	730	0.5	0.3
Vessel Inlet Nozzle	0.375	730	2.5	1.6
IHX Outlet Nozzle	0.375	730	2.5	1.6
IHX Shell	1.625	730	5.9	3.7
Isolation Valve Inlet Nozzle	0.375	730	2.5	1.6
Isolation Valve Outlet Nozzle	0.375	730	2.5	1.6
Isolation Valve Body	1.0	730	1.0	0.6
Piping	0.5	730	1.9	1.2
<u>Intermediate System Cold Leg</u>				
IHX Inlet Nozzle	0.375	651	1.3	1.2
IHX Shell	0.375	651	1.3	1.2
Piping	0.375	651	1.3	1.2
Isolation Valve Inlet Nozzle	0.375	651	1.3	1.2
Isolation Valve Outlet Nozzle	0.375	651	1.3	1.2
Isolation Valve Body	0.75	651	0.7	0.6
Pump Suction Nozzle	0.375	651	1.3	1.2
Pump Discharge Nozzle	0.375	651	1.3	1.2
Pump Casing	2.0	651	0.3	0.2
Steam Generator Outlet Nozzle	0.375	651	-	-
Steam Generator Shell	0.5	651	-	-
Steam Generator Tubing	0.109	651	-	-

- (a) The interstitial diffusion distance is arbitrarily taken to be the depth at which the (C + N) concentration has changed from  $C_0$  to within 90 percent of the starting (C + N) concentration.
- (b) Type 304 was assumed to have 0.06 weight percent carbon and 0.06 weight percent nitrogen. Type 304 L was assumed to have 0.02 weight percent carbon and 0.02 weight percent nitrogen.

TABLE 15

Costs for Stainless Steel Plate<sup>(73)</sup>

MATERIAL	COST, 100 LBS. (\$)
304 SS	52.00
304 L SS	58.00
316 SS	79.00
316 L SS	85.00



TABLE 16

Summary of Austenitic Stainless Steel Performance Under LMFBR Conditions

PROPERTY	ALLOY RATING	COMMENTS
Mechanical Properties	1. 316 H SS 2. 316 SS 3. 304 H SS 4. 304 SS	For maximum strength Type 316 grades generally are superior. This is especially true for creep-rupture strength. Under short-term tensile testing Type 316 has superior ductility.
Fabrication and Weldability	1. 304 SS 2. 316 SS 3. 304 H SS 4. 316 H SS	There may be difficulty in avoiding sensitization during stress relief and welding of high carbon grades. Type 304 has better weldability than Type 316.
Sensitization	1. 304 SS 2. 316 SS 3. 304 H SS 4. 316 H SS	Type 304 resists sensitization more effectively than Type 316. High carbon grades are likely to be sensitized at some stage of welding, fabrication or stress relief.
Stress-Corrosion Cracking	1. 316 SS 2. 316 H SS 3. 304 SS 4. 304 H SS	The higher strength of Type 316 appears to be an advantage in chloride-induced cracking. Carbon is detrimental at high (> 0.1 wt. percent) levels. Under caustic contamination conditions the relative behavior of Type 304 and 316 is unknown.
Sodium Corrosion	1. 304 SS, 316 SS, 304 H SS, 316 H SS	There should be little difference in the corrosion rates of these alloys.
Interstitial Transfer	1. 316 SS 2. 316 H SS 3. 304 SS 4. 304 H SS	Interstitial loss from hot-leg regions, and interstitial absorption in cold-leg regions, is minimized in Type 316 grades of steel.

TABLE 17(a)

## Tentative Material Selections for CHBRP Hot-Leg Components

COMPONENT	RECOMMENDED MATERIAL	REASON FOR SELECTION
<u>Primary System</u>		
Vessel	304 SS	Adequate strength, lower cost, ease of fabricability and weldability and greater resistance, compared to 316 SS, to sensitization during fabrication and heat treatment.
Pump	304 SS	
Isolation Valve	304 SS	
Vessel Outlet Nozzle	316 SS	Superior strength compared to 304 SS and better resistance to interstitial transfer.
IHX Inlet Nozzle	316 SS	
Piping	316 H SS	Has high mechanical strength. Specification of H grade will guarantee carbon level $\geq 0.04$ w/o.
<u>Intermediate System</u>		
Isolation Valve	304 SS	As above for primary system valve.
IHX Shell	304 SS	As above for primary vessel.
IHX Outlet Nozzle	316 SS	As above for primary vessel outlet nozzle. Type 304 SS may be used if maximum strength not mandatory.
Piping	316 SS	The lower operating temperature, compared to the primary hot-leg piping, should allow regular, rather than high-carbon, material to be used. This will reduce sensitization effects, but maintain resistance to interstitial transfer.
<u>Auxiliary System</u>		
Coolant Overflow and Return Piping	316 H SS	The thinner walls require maximum resistance to interstitial losses. H grade guarantees carbon level $\geq 0.04$ w/o.
Primary Sample Loop Piping	316 H SS	As above for coolant overflow piping.
Isolation Valve	304 SS	As above for primary system valve.

TABLE 17(b)

Tentative Material Selections for CRBRP Cold-Leg Components

COMPONENT	RECOMMENDED MATERIAL	REASON FOR SELECTION
<u>Primary System</u>		
Vessel	304 SS	See Table 17(a)
IHX Shell	304 SS	
Isolation Valve	304 SS	
Vessel Inlet Nozzle	304 SS	Adequate strength, lower cost, ease of fabricability and weldability, and greater resistance, compared to 316 SS, to sensitization during fabrication and heat treatment.
IHX Outlet Nozzle	304 SS	
Piping	316 SS	Superior strength compared to Type 304 and 304 H (See Table 2) and increased resistance to carburization and nitridation effects.
<u>Intermediate System</u>		
Isolation Valve	304 SS	As above for primary system valve.
Pump	304 SS	
IHX Inlet Nozzle	304 SS	
Piping	316 SS*	As above from primary system piping.

\* Possible use of 2-1/4 Cr-1 Mo if cost is less than that for Type 316 stainless steel

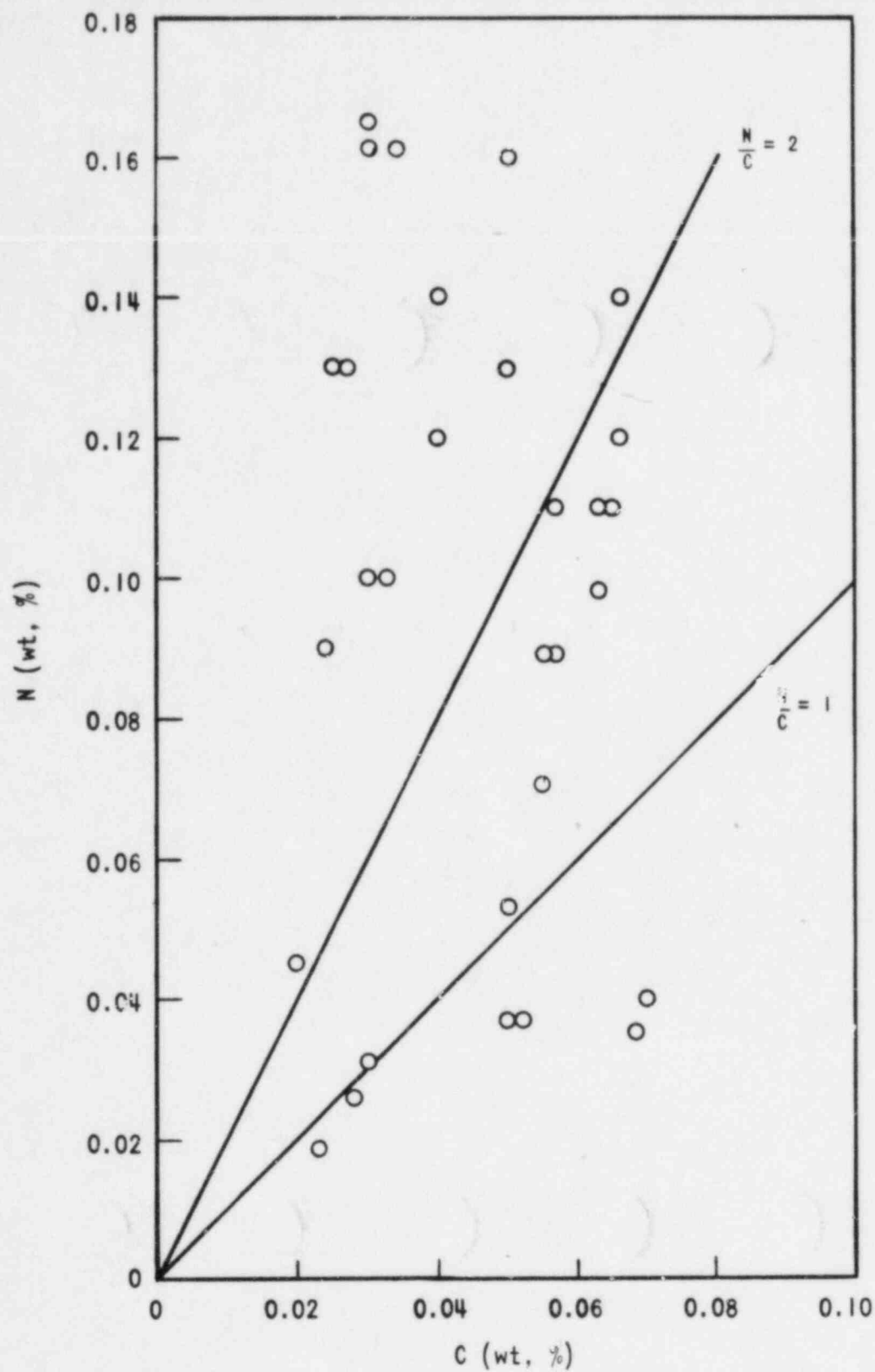


Figure 1. Nitrogen/Carbon Ratios From Type 304 Stainless Steel Literature Review

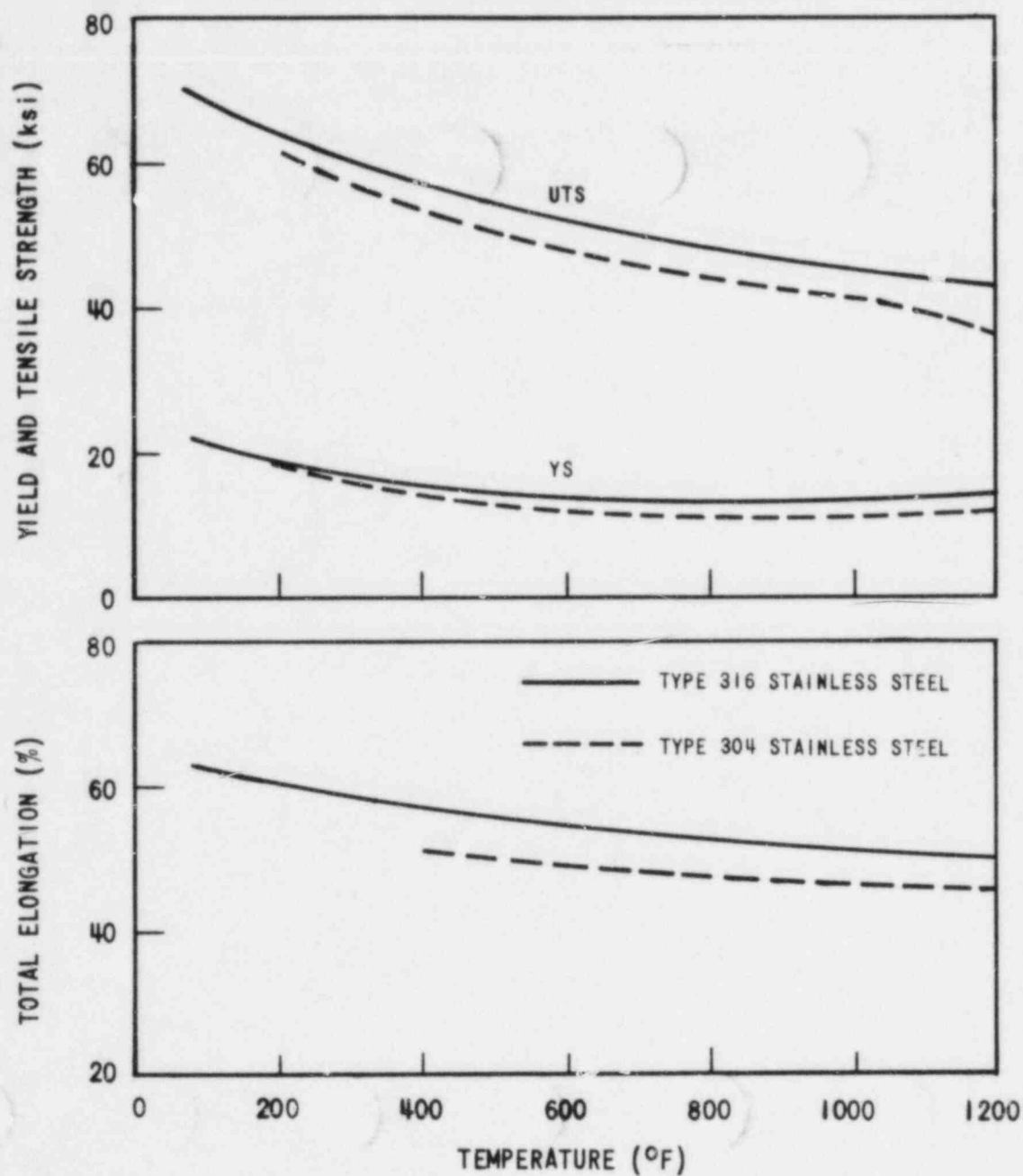


Figure 2. Comparison Between the Strengths and Ductilities of Solution Treated Types 304 and 316 Stainless Steel Containing 0 Weight Percent (C + N)

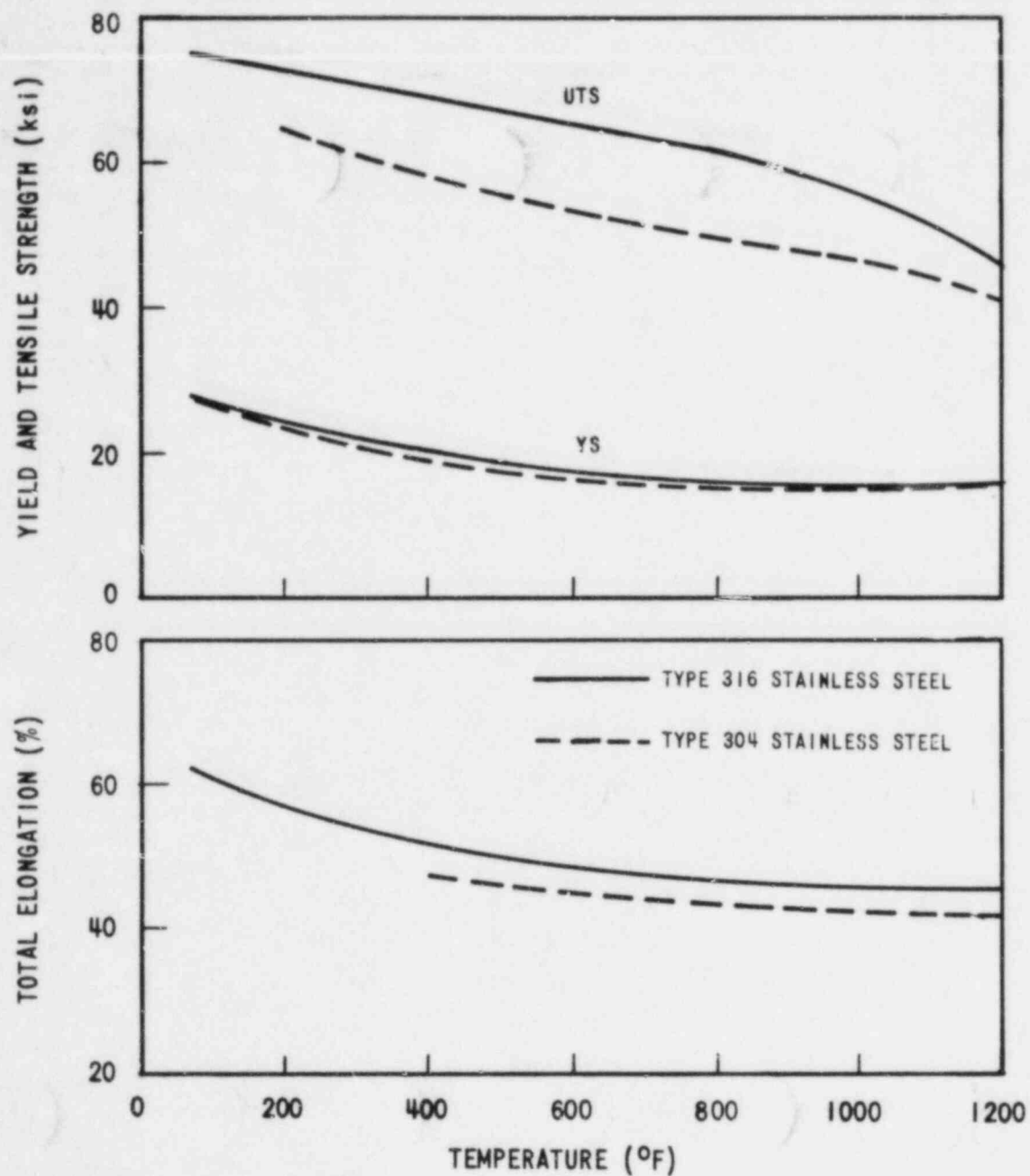


Figure 3: Comparison Between the Strengths and Ductilities of Solution Treated Types 304 and 316 Stainless Steel Containing 0.04 Weight Percent (C + N)

6129-3



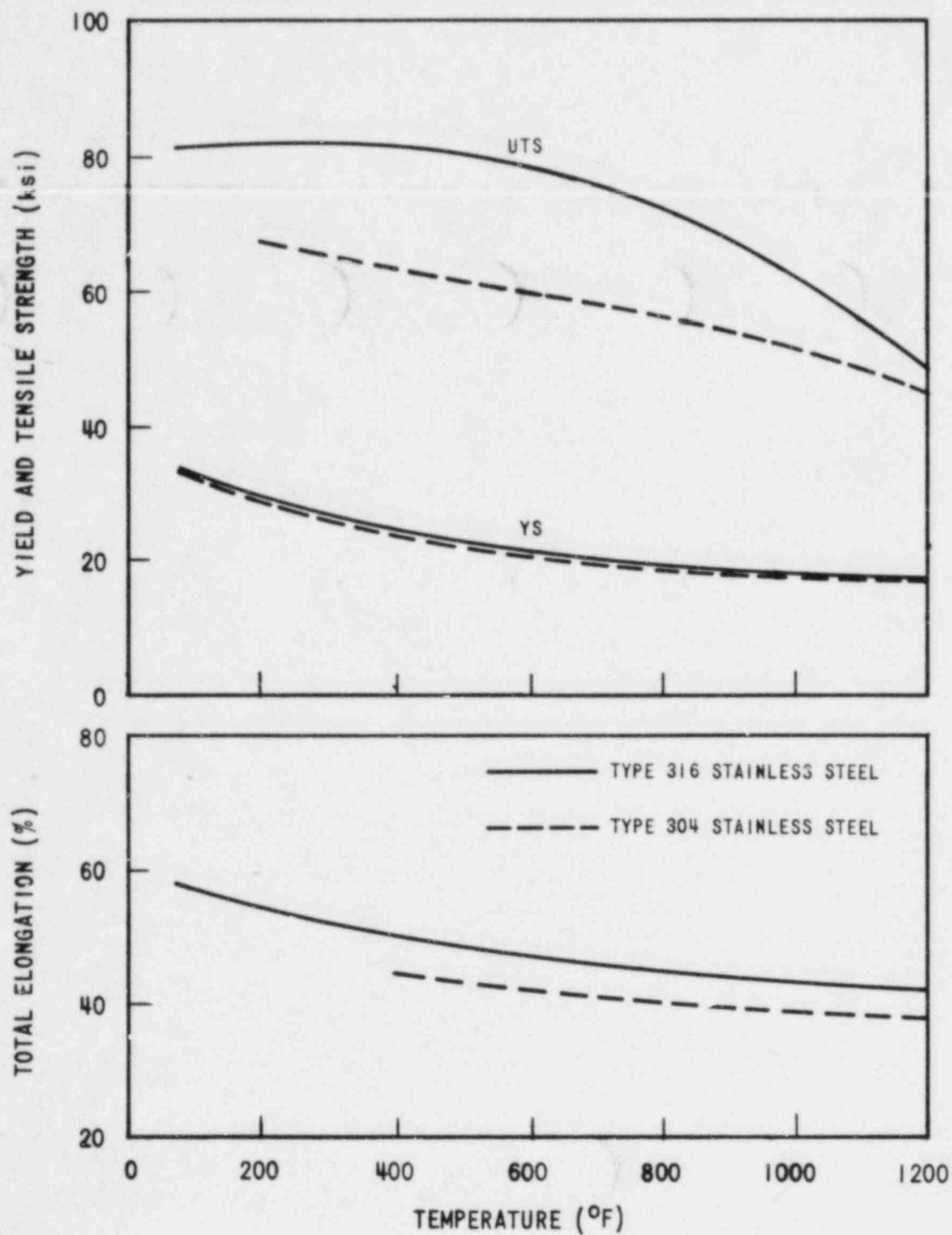


Figure 4. Comparison Between the Strengths and Ductilities of Solution Treated Types 304 and 316 Stainless Steel Containing 0.08 Weight Percent (C + N)

6129-4

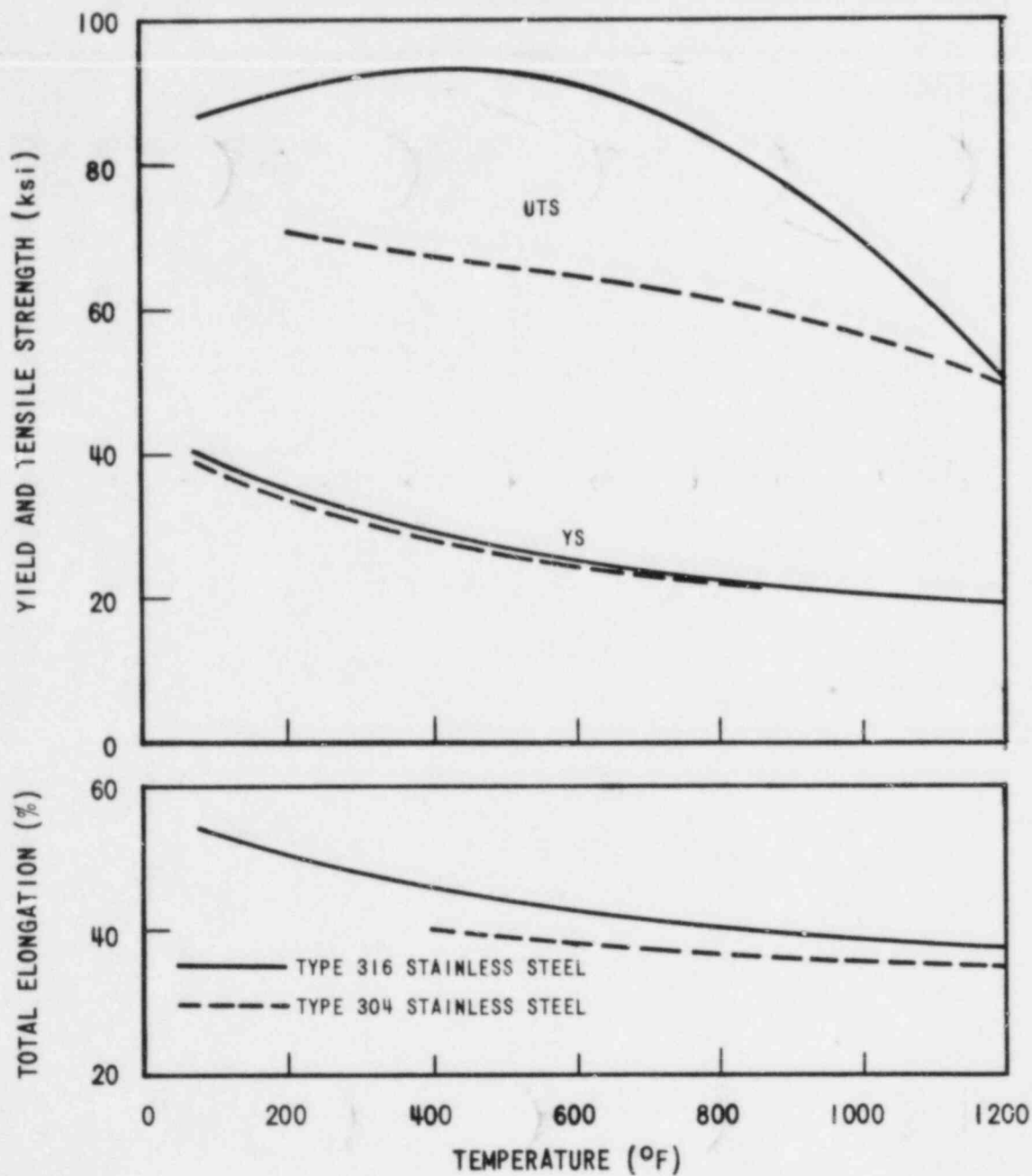


Figure 5. Comparison Between the Strengths and Ductilities of Solution Treated Types 304 and 316 Stainless Steel Containing 0.12 Weight Percent (C + N)

6129-5

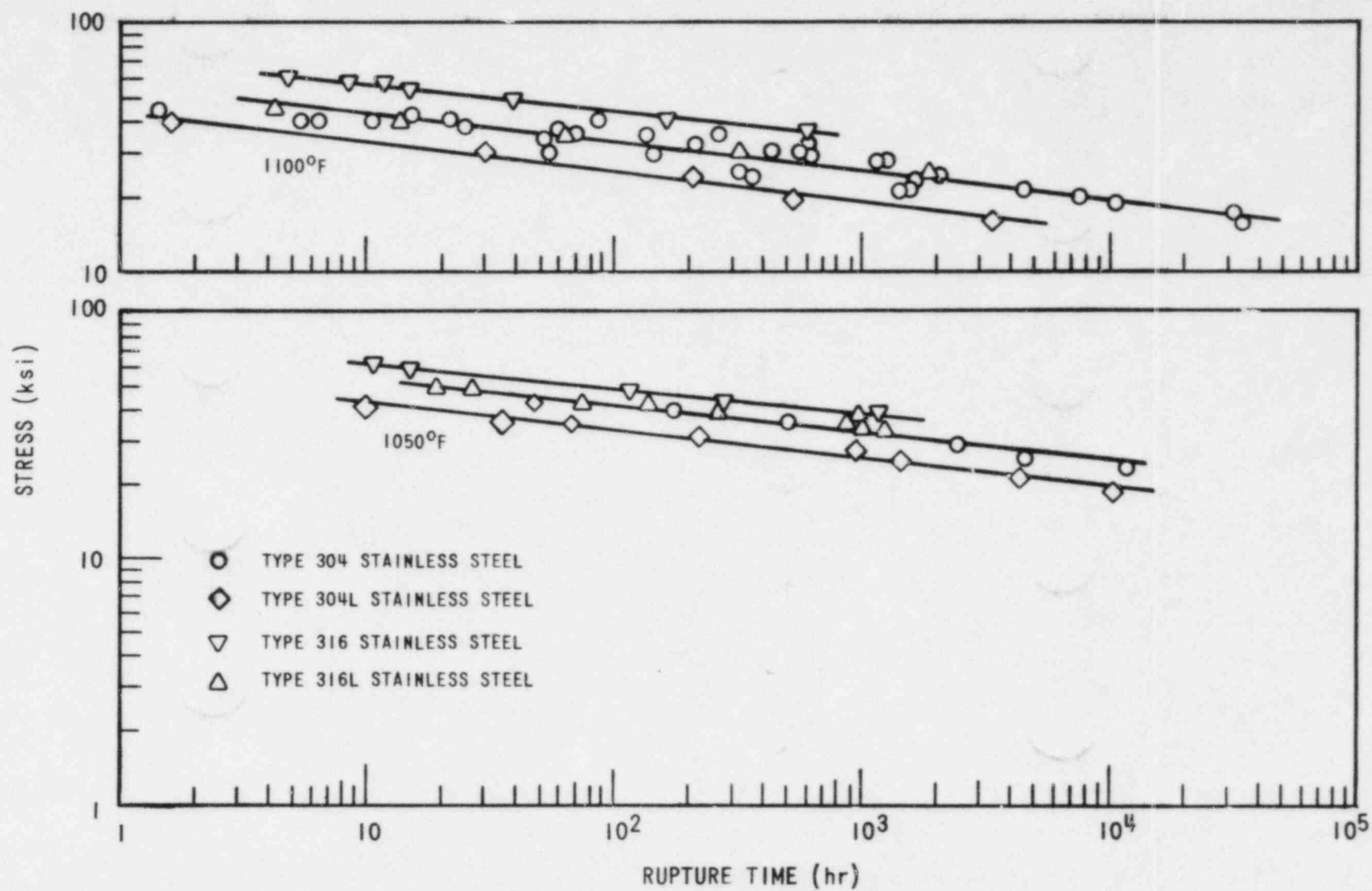


Figure 6. Stress-Rupture for Austenitic Stainless Steel Bar and Plate

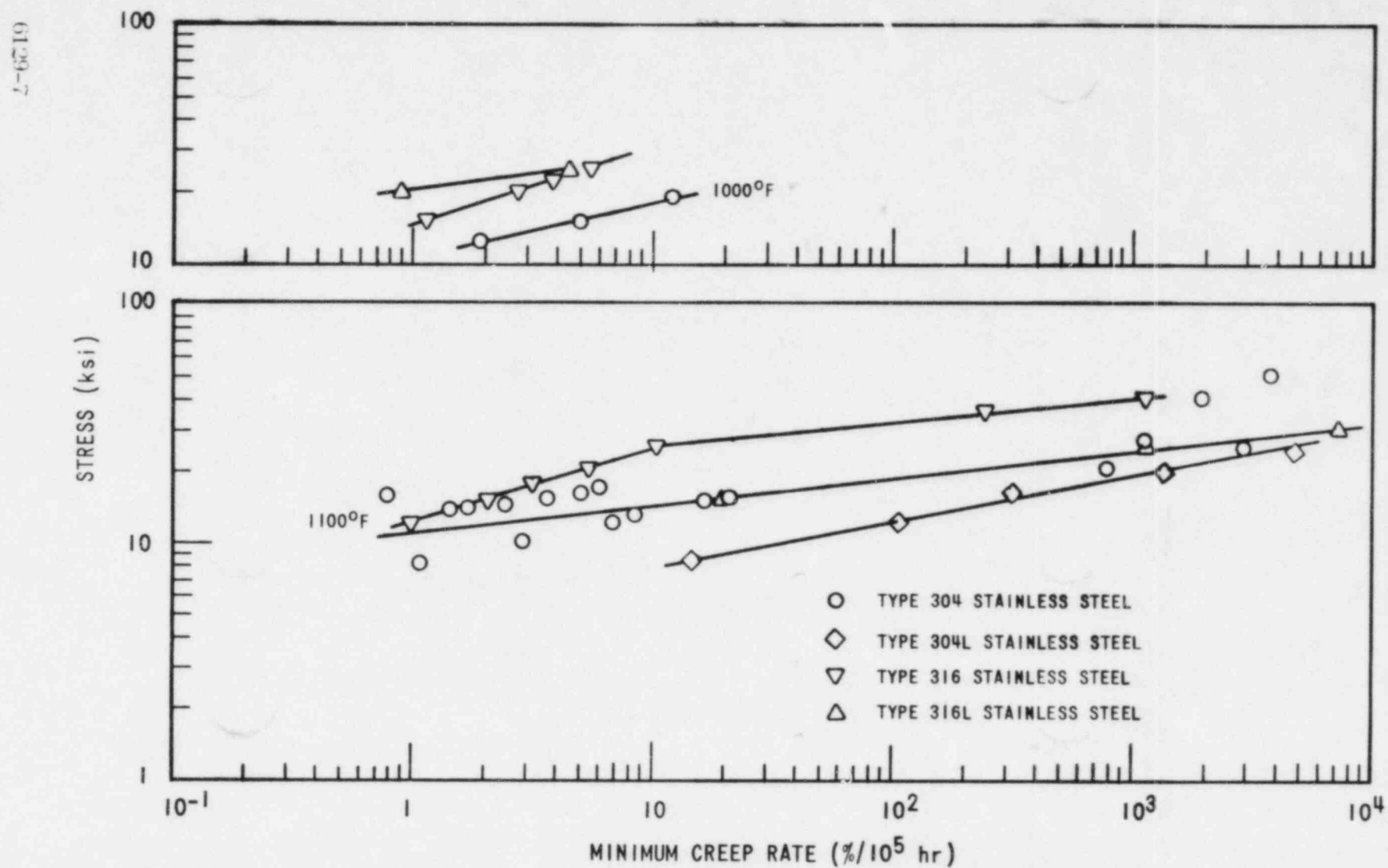


Figure 7. Minimum Creep Rates for Austenitic Stainless Steel Bar and Plate

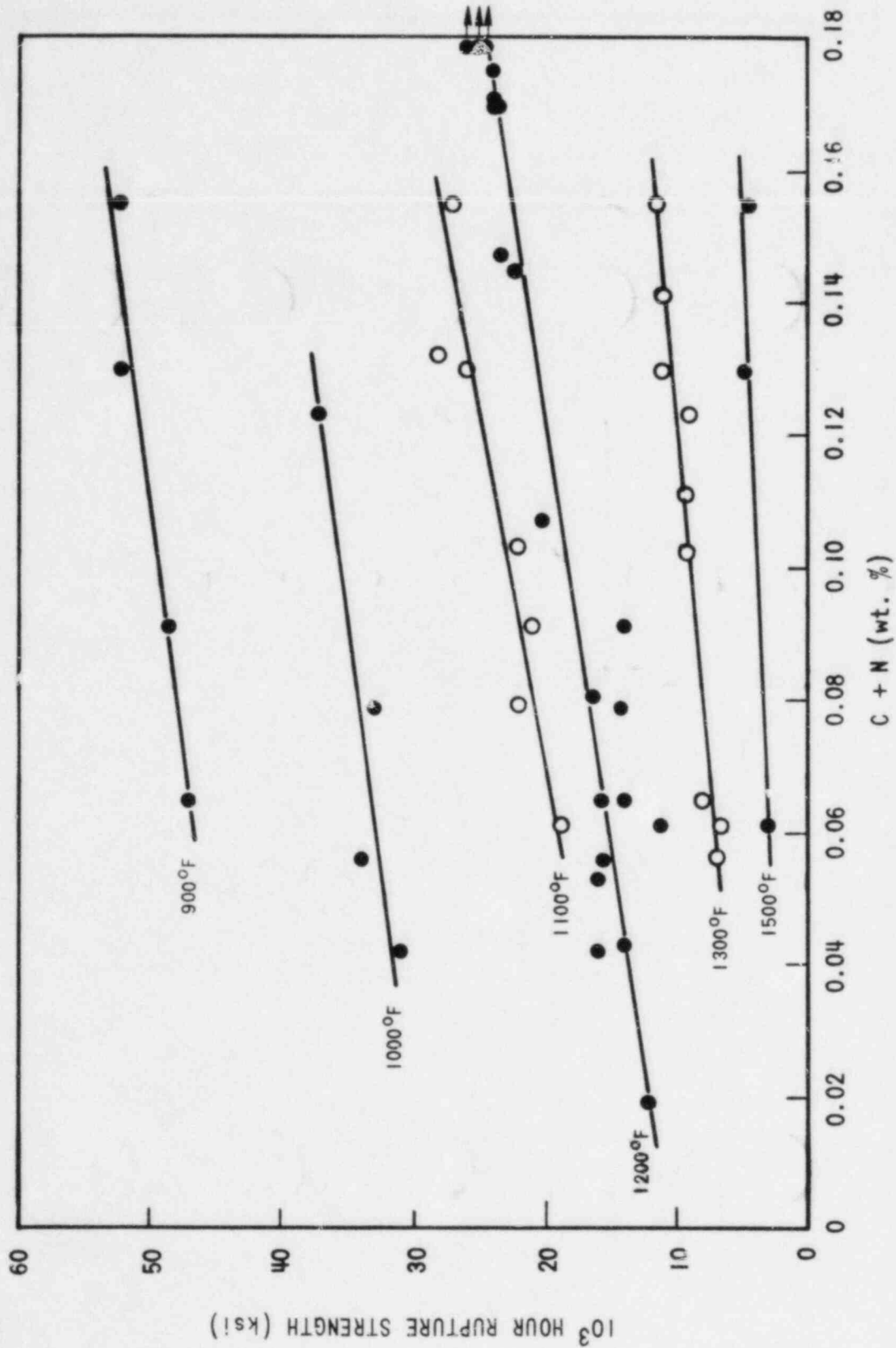


Figure 8. Effect of Interstitial Content on the  $10^3$  Hour Rupture Strength of Type 304 Stainless Steel Bar

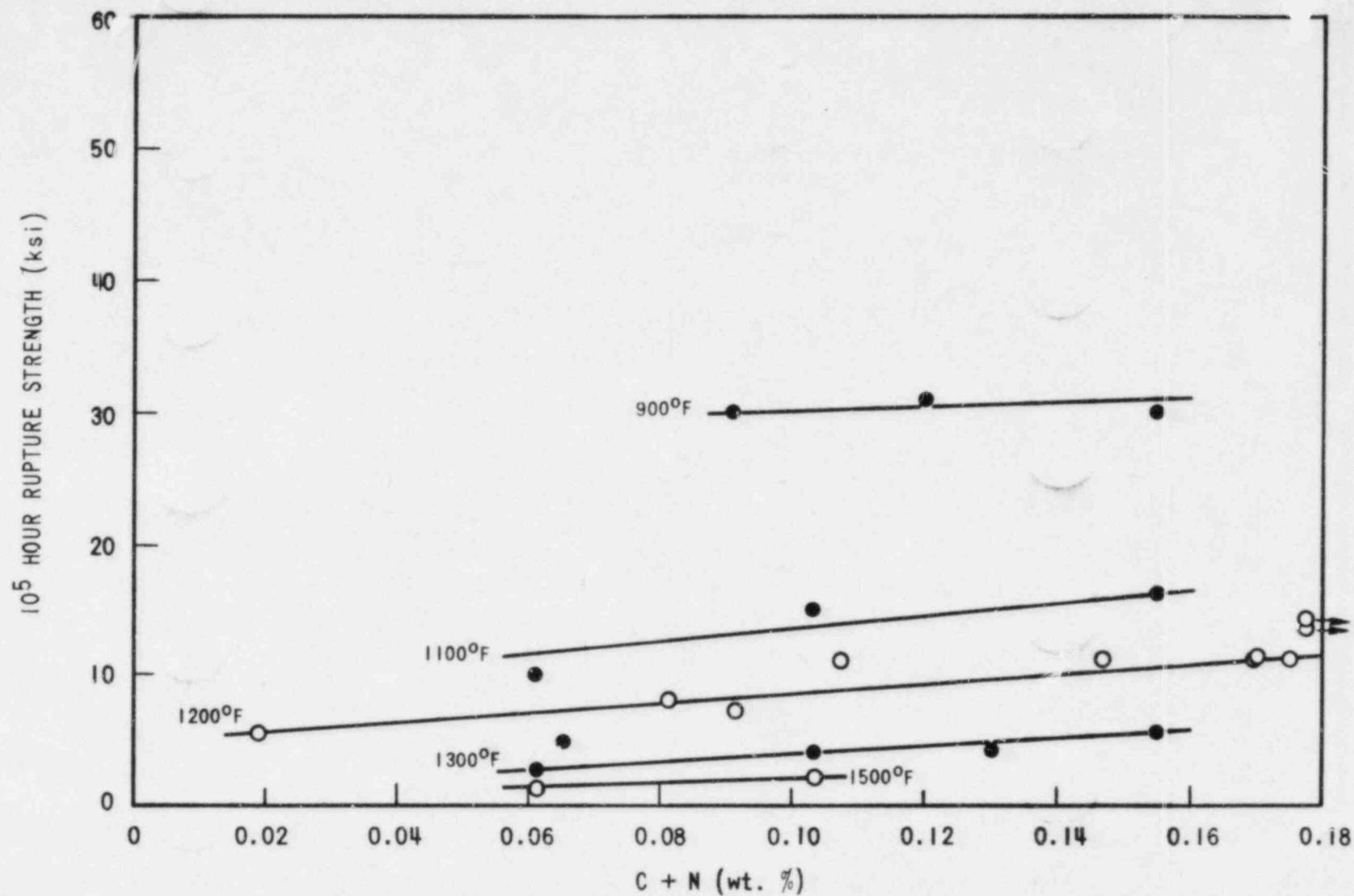


Figure 9. Effect of Interstitial Content on the 10<sup>5</sup> Hour Rupture Strength of Type 304 Stainless Steel Bar



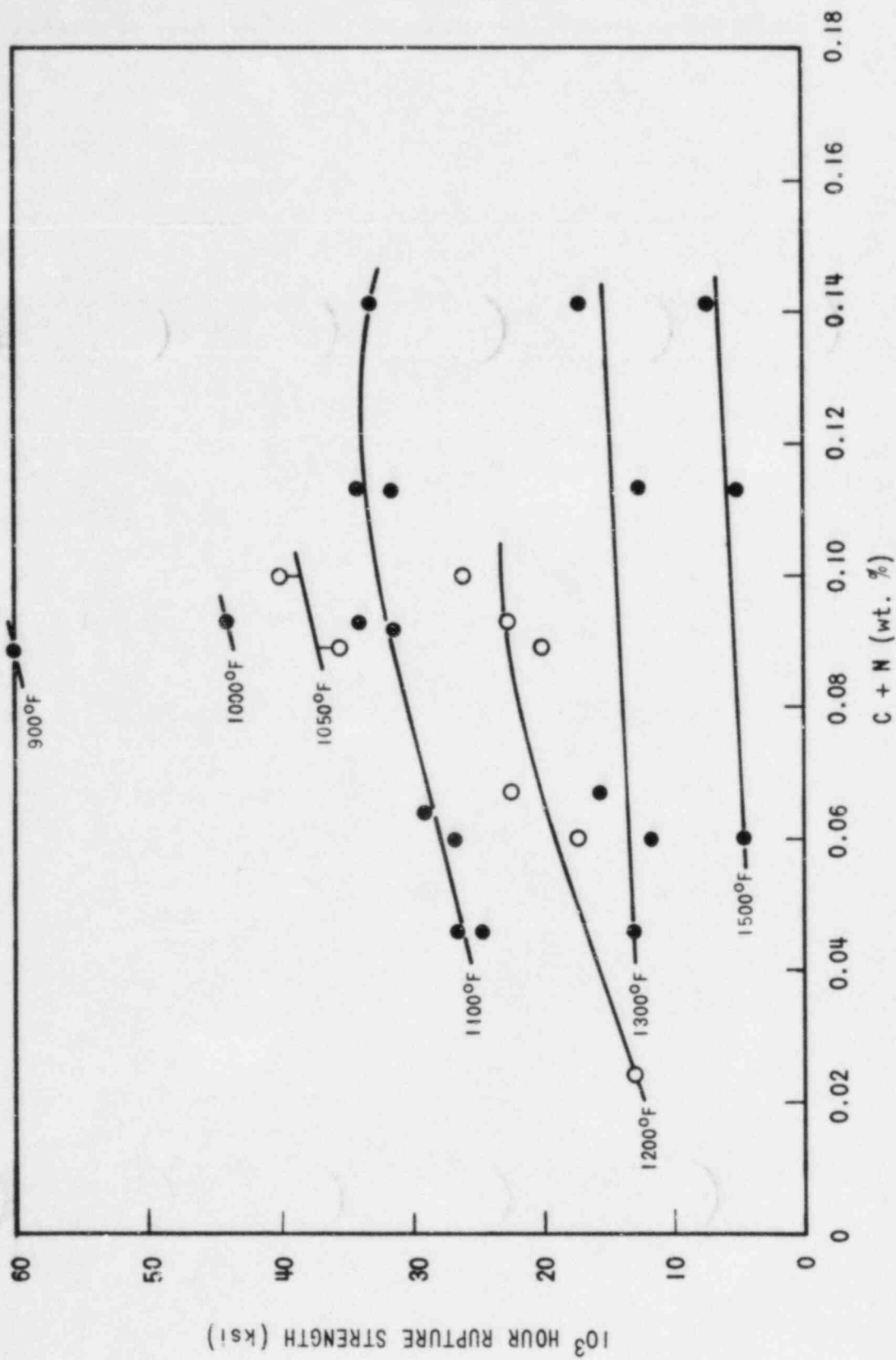


Figure 10. Effect of Interstitial Content on the 10<sup>3</sup> Hour Rupture Strength of Type 316 Stainless Steel Bar

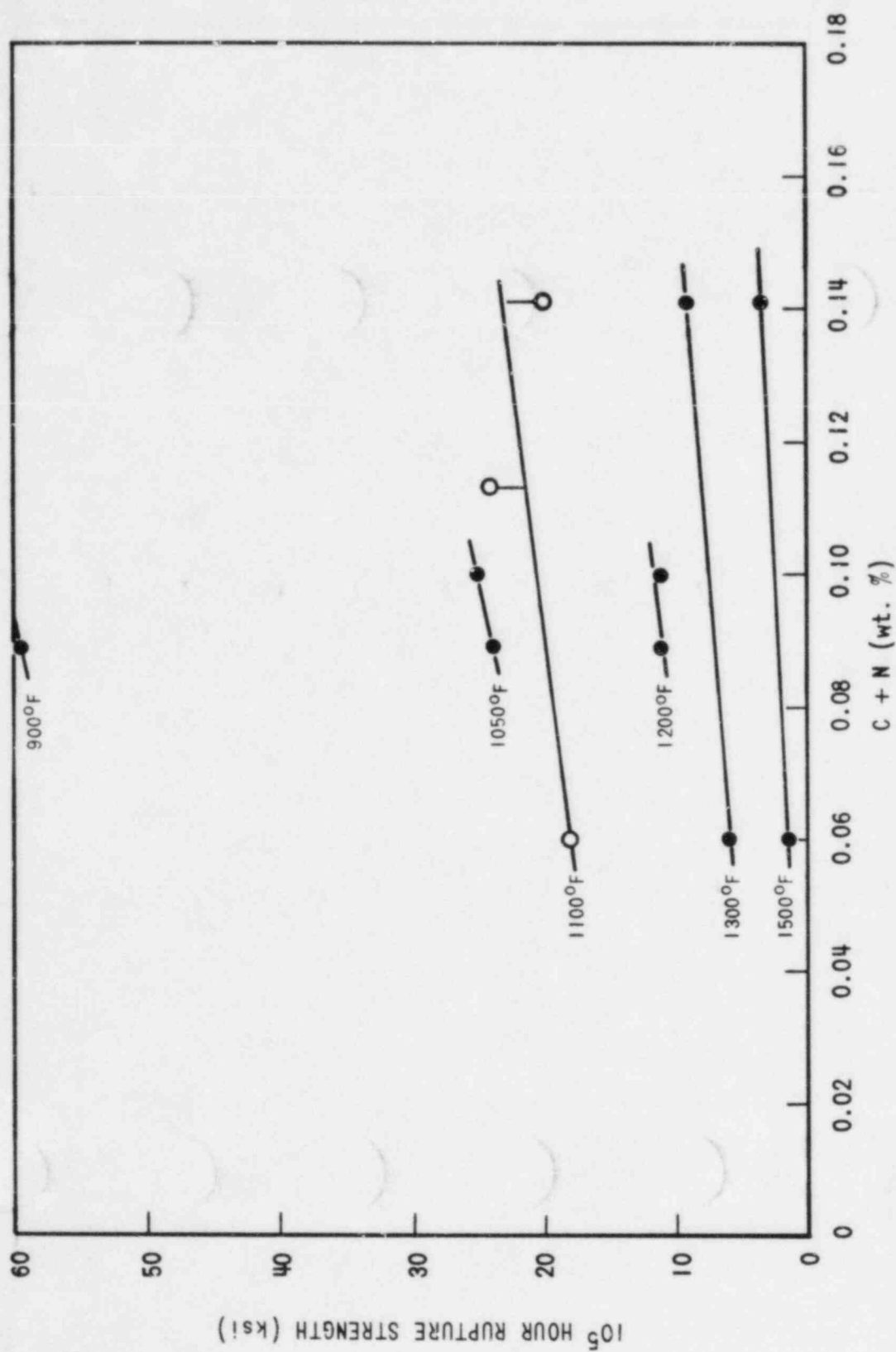


Figure 11. Effect of Interstitial Content on the 10<sup>5</sup> Hour Rupture Strength of Type 316 Stainless Steel Bar

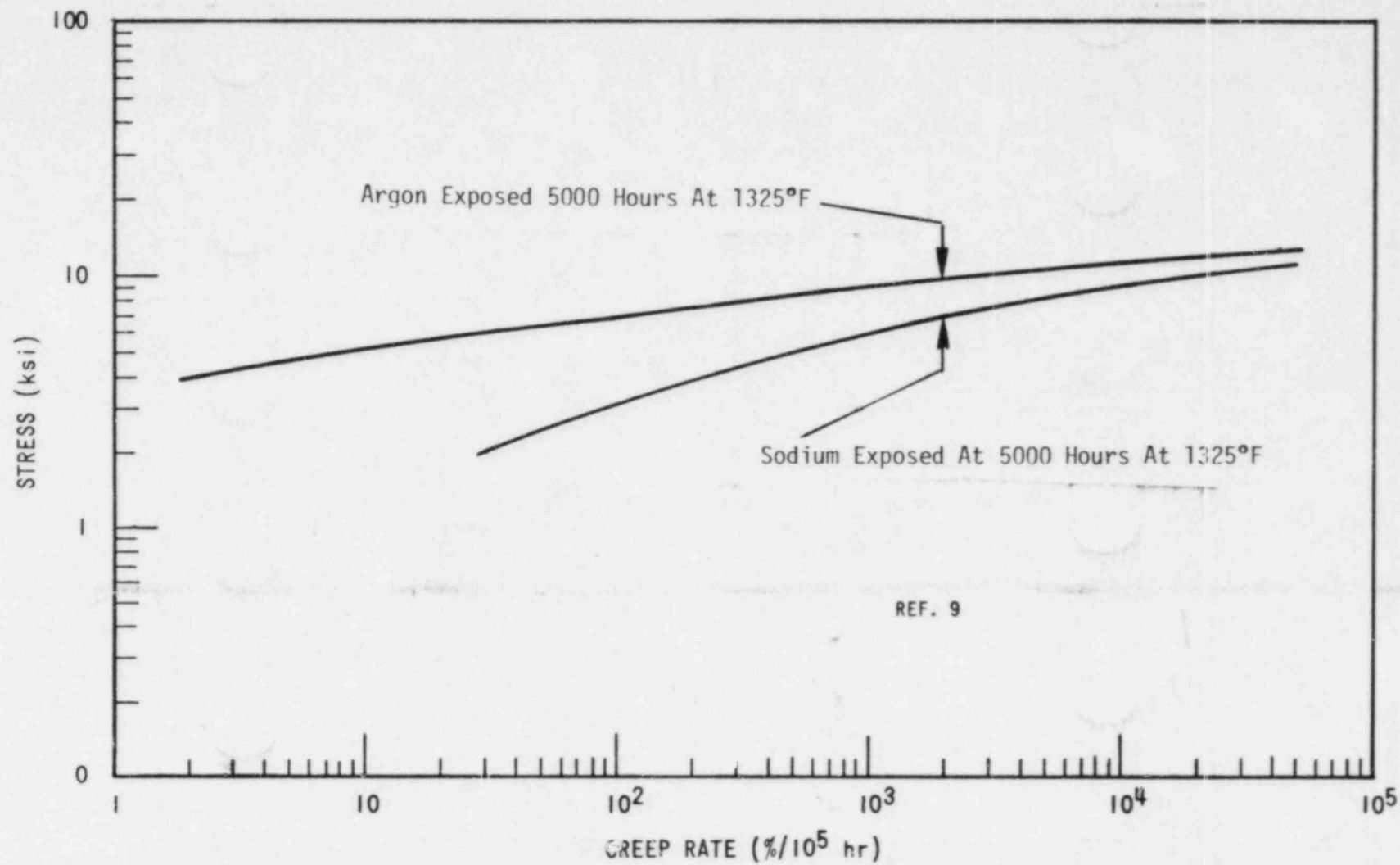


Figure 12. Effect of Sodium Exposure on the Creep Rates of Type 316 Stainless Steel at 1325°F

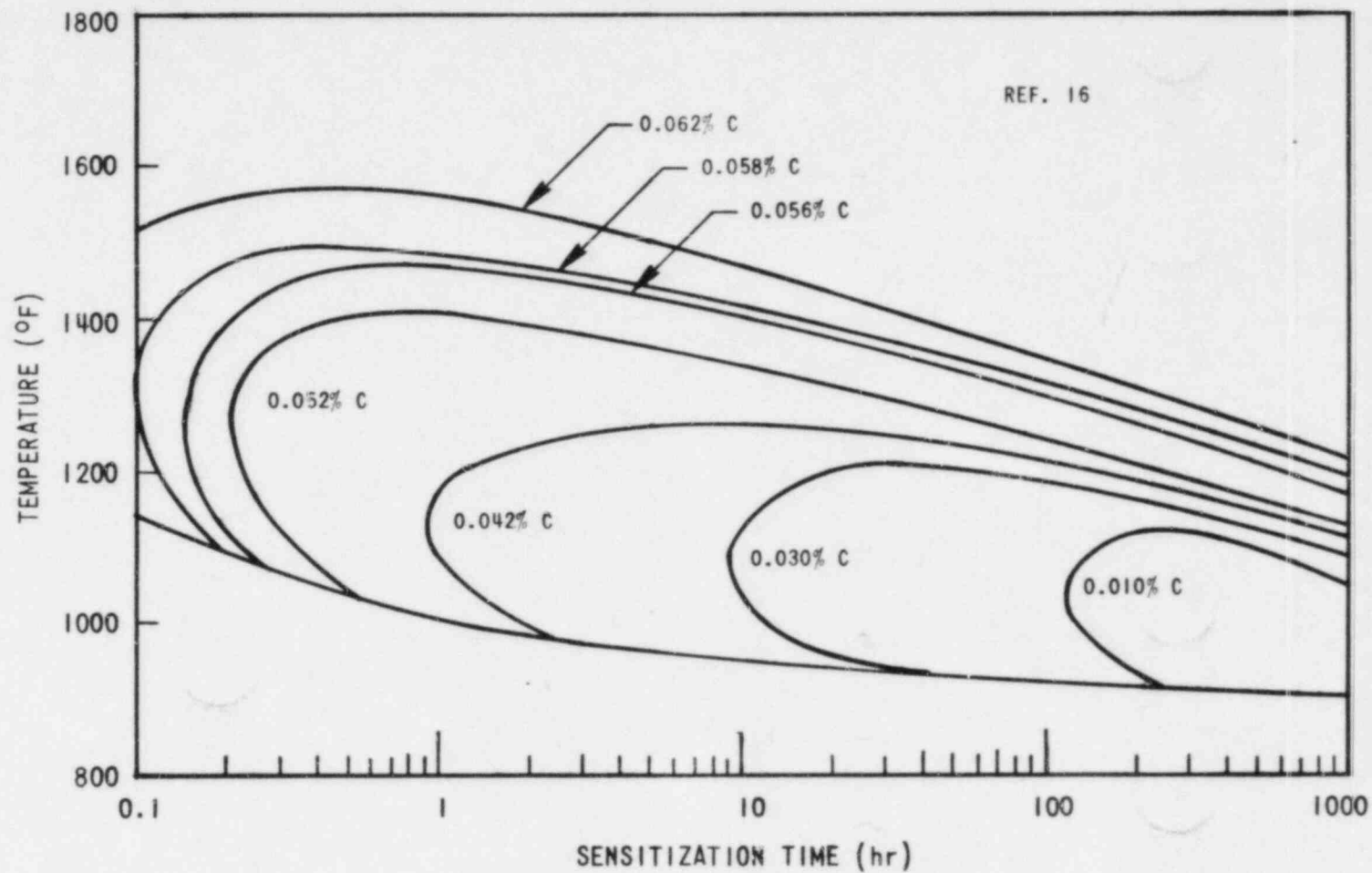


Figure 13. Onset of Sensitization in 18-8 Stainless Steel as a Function of Carbon as Determined by the Strauss Test

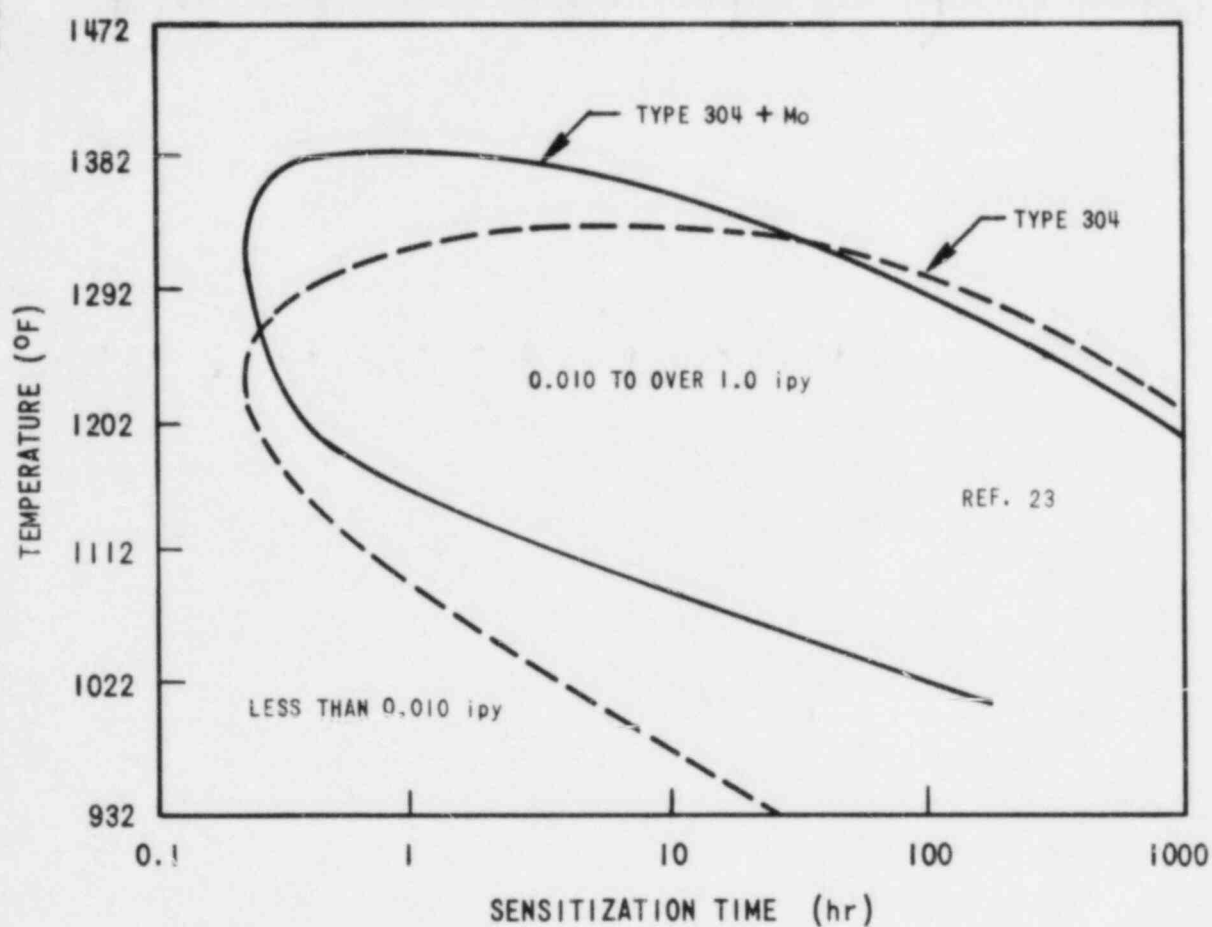


Figure 14. Effect of Molybdenum on the Time-Temperature-Sensitization Diagram of Type 304 (0.054% C) Stainless Steel

6129-14

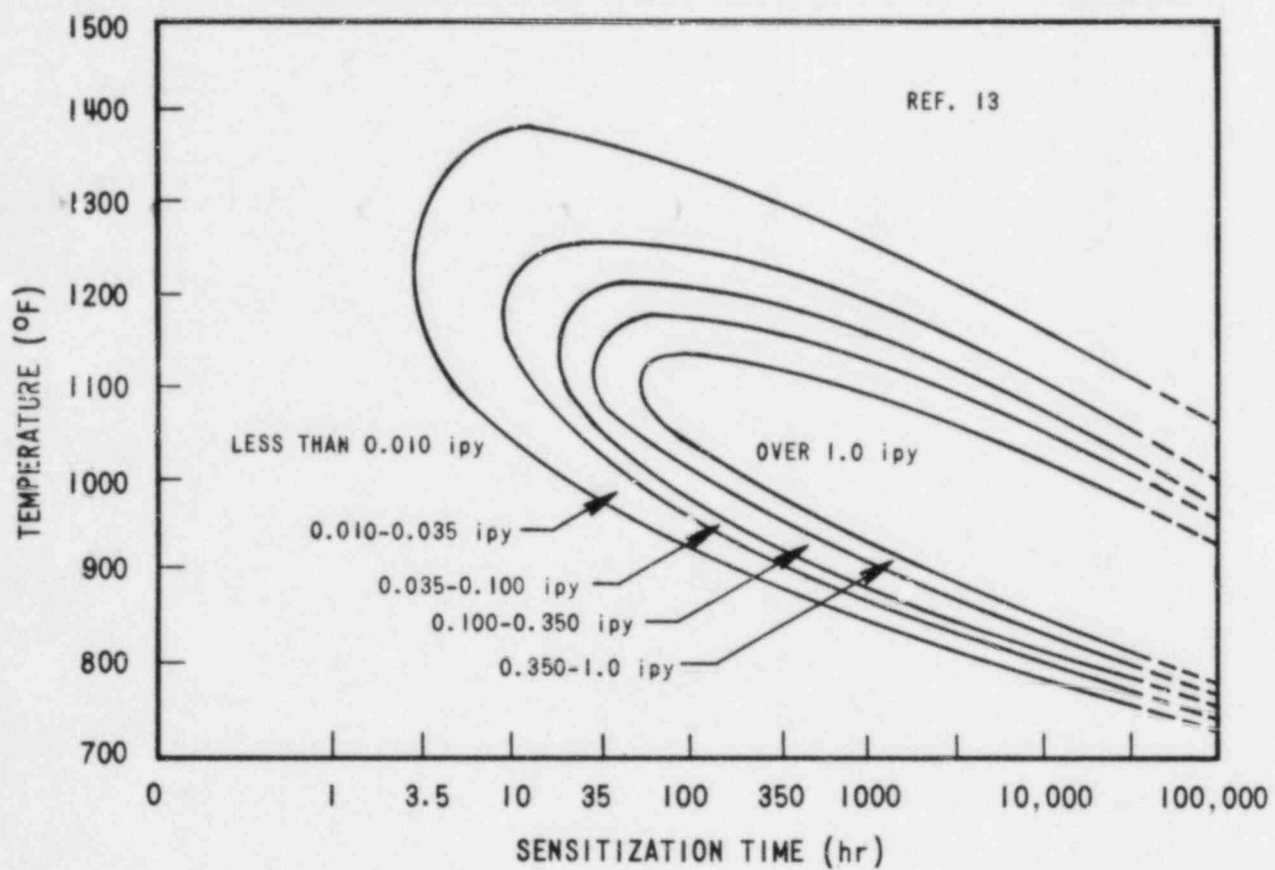


Figure 15. Time-Temperature-Sensitization Diagram of Type 304 L Stainless Steel (Huey Test)

6129-15

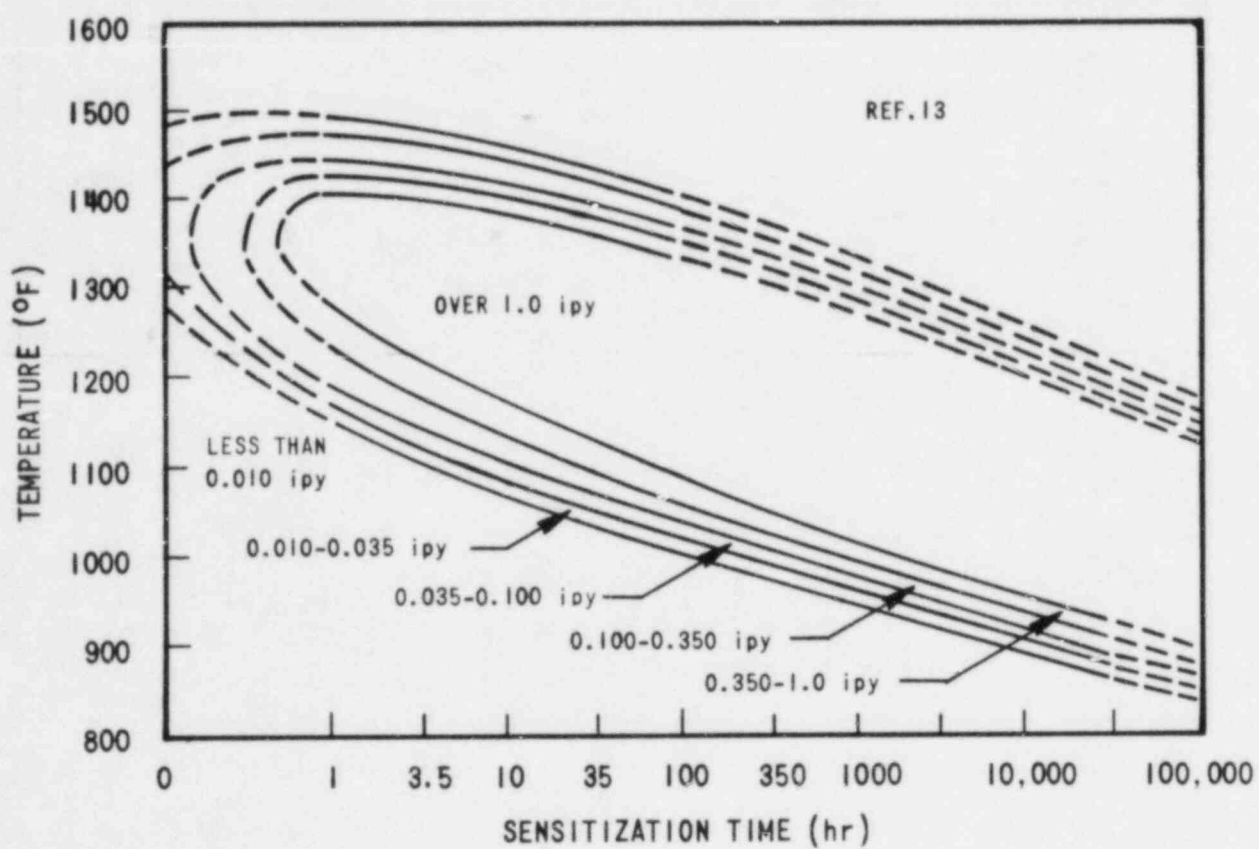


Figure 16. Time-Temperature-Sensitization Diagram of Type 316 L Stainless Steel (Huey Test)

6129-16



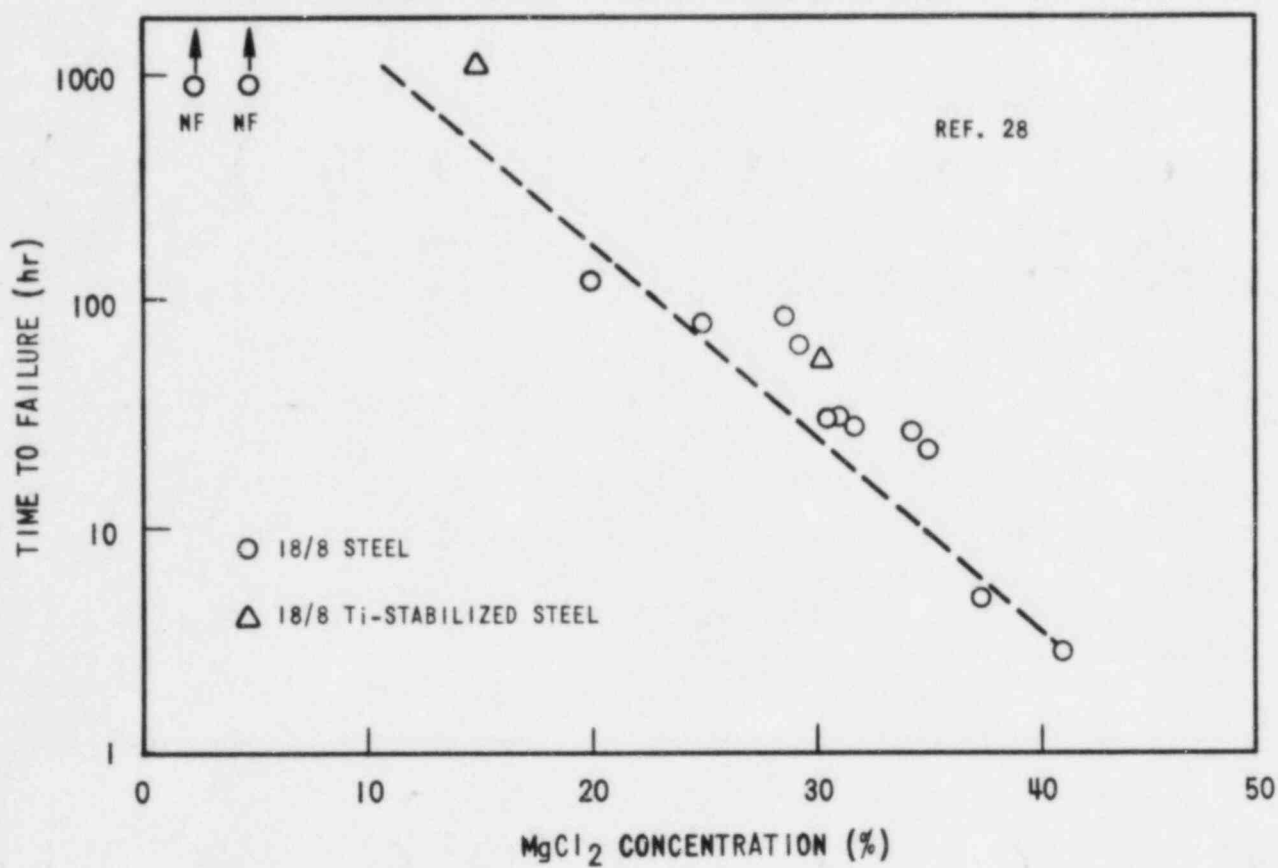


Figure 17. Effect of MgCl<sub>2</sub> Concentration on Time to Failure

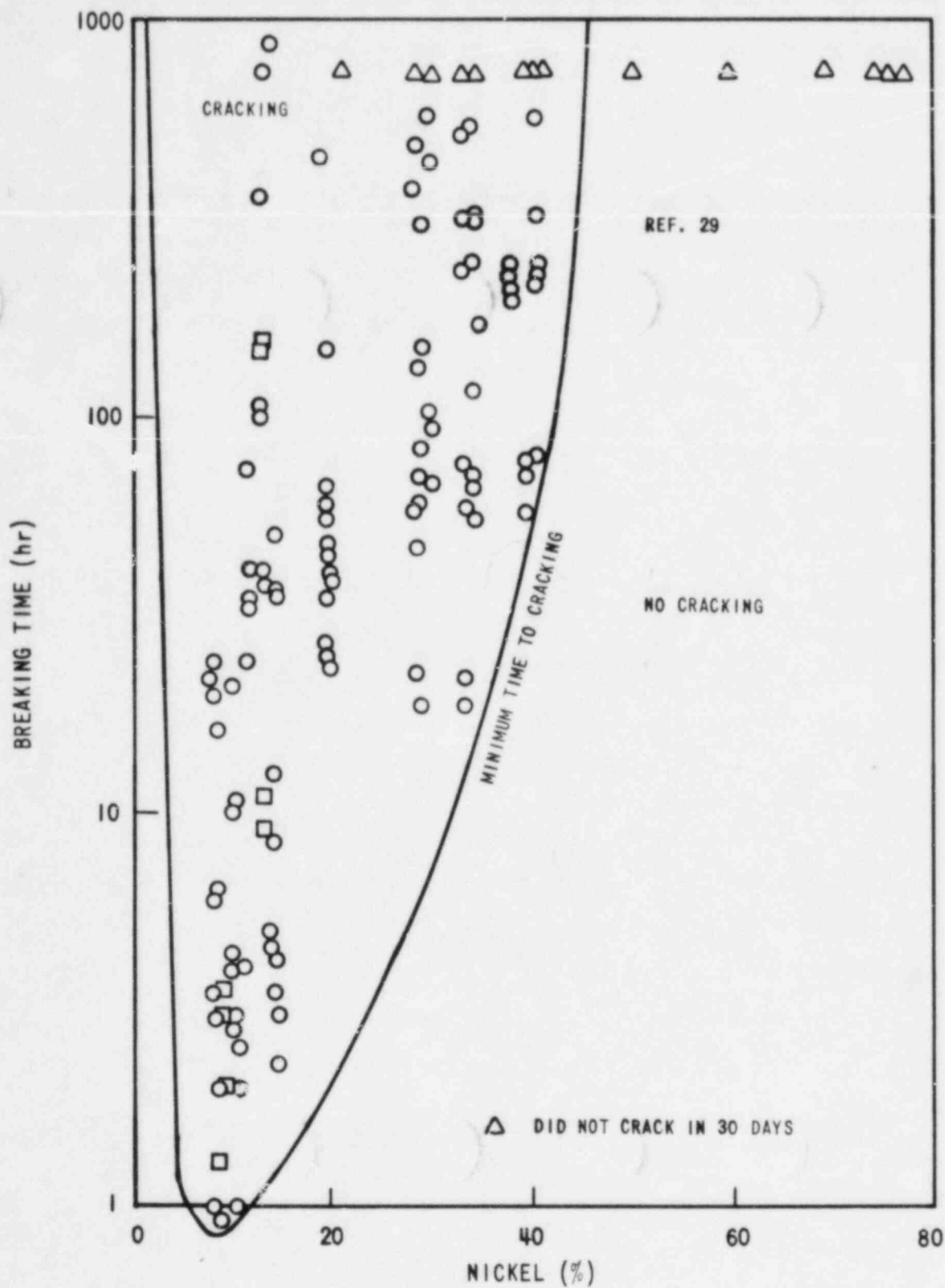


Figure 18. Effect of Increasing the Nickel Content on the Susceptibility of Fe-Cr-Ni Wires in Boiling 42% MgCl<sub>2</sub>

6129-18

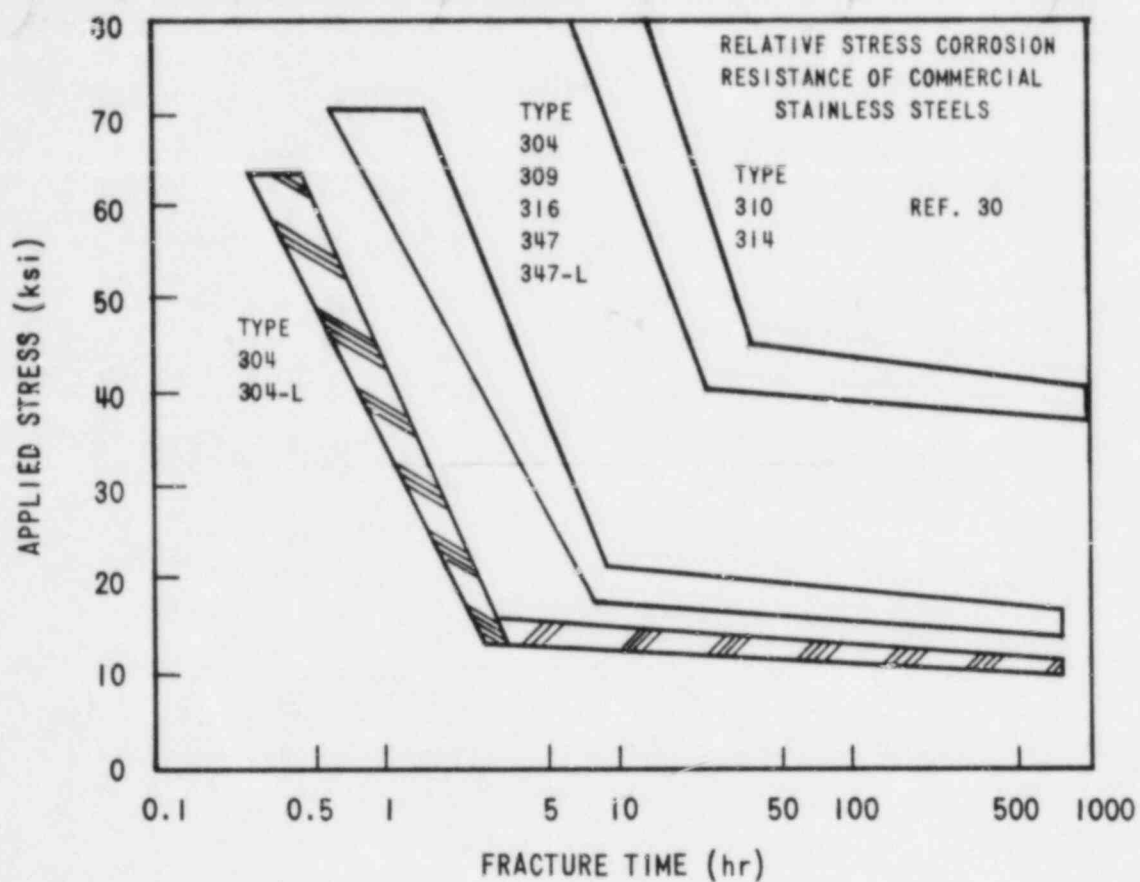


Figure 19. Composite Curves Illustrating the Relative Stress Corrosion Cracking Resistance for Commercial Stainless Steels in Boiling 42 Percent Magnesium Chloride

6129-19

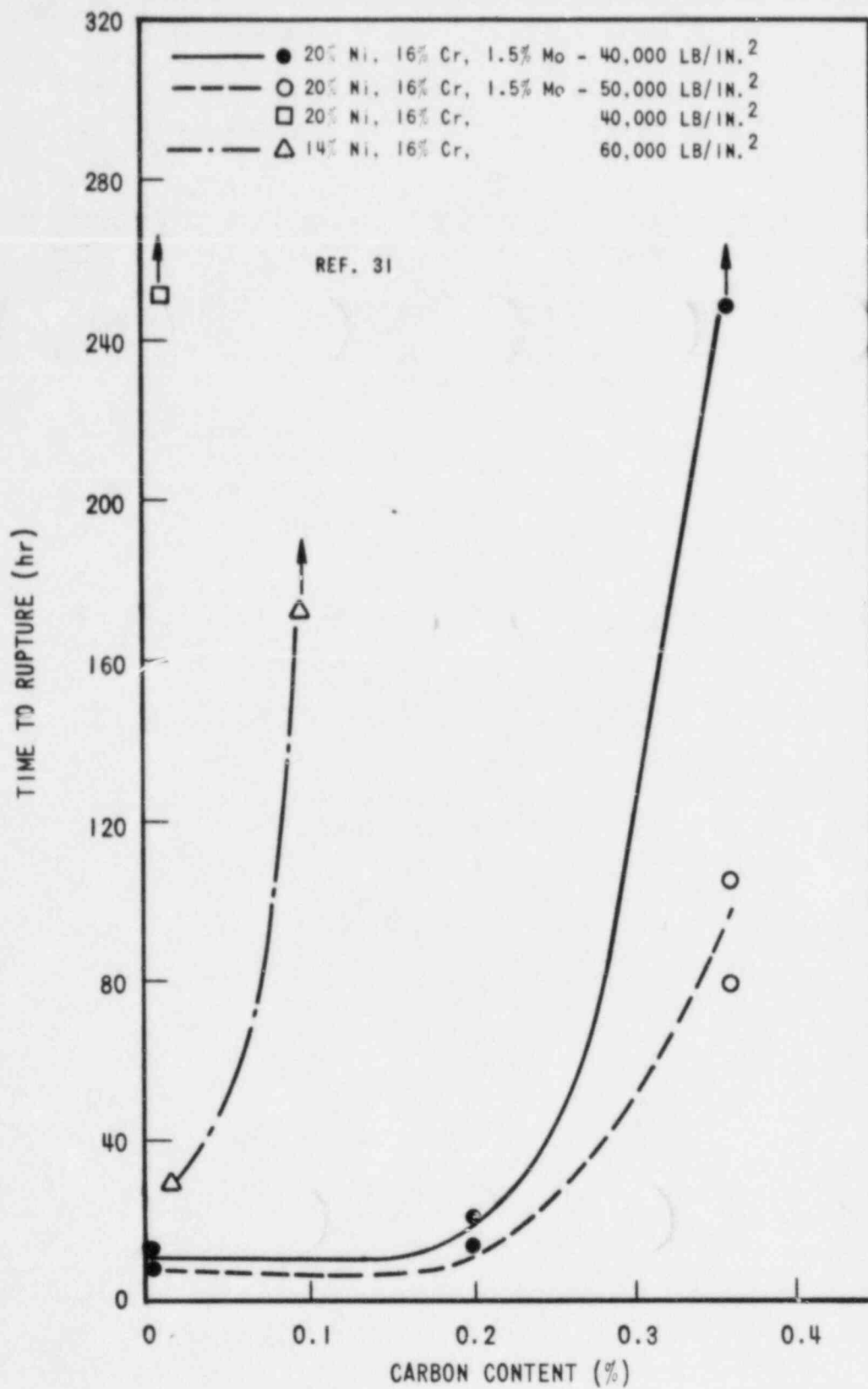


Figure 20. Effect of Carbon on Times to Failure of Stainless Steel in Boiling  $\text{MgCl}_2$

6129-20

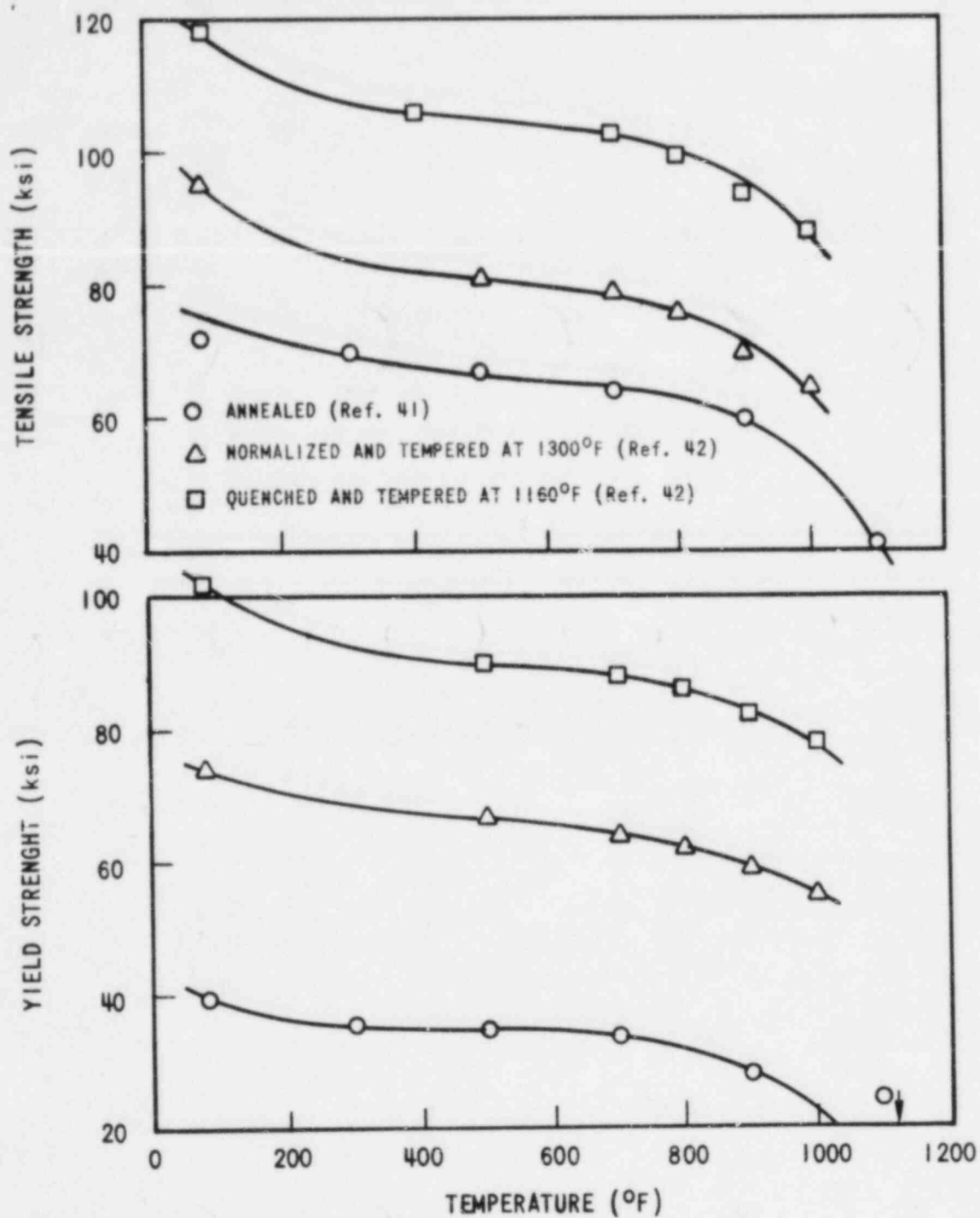


Figure 21. Effect of Heat Treatment on the Yield and Tensile Strengths of 2 1/4 Cr-1 Mo

6129-21

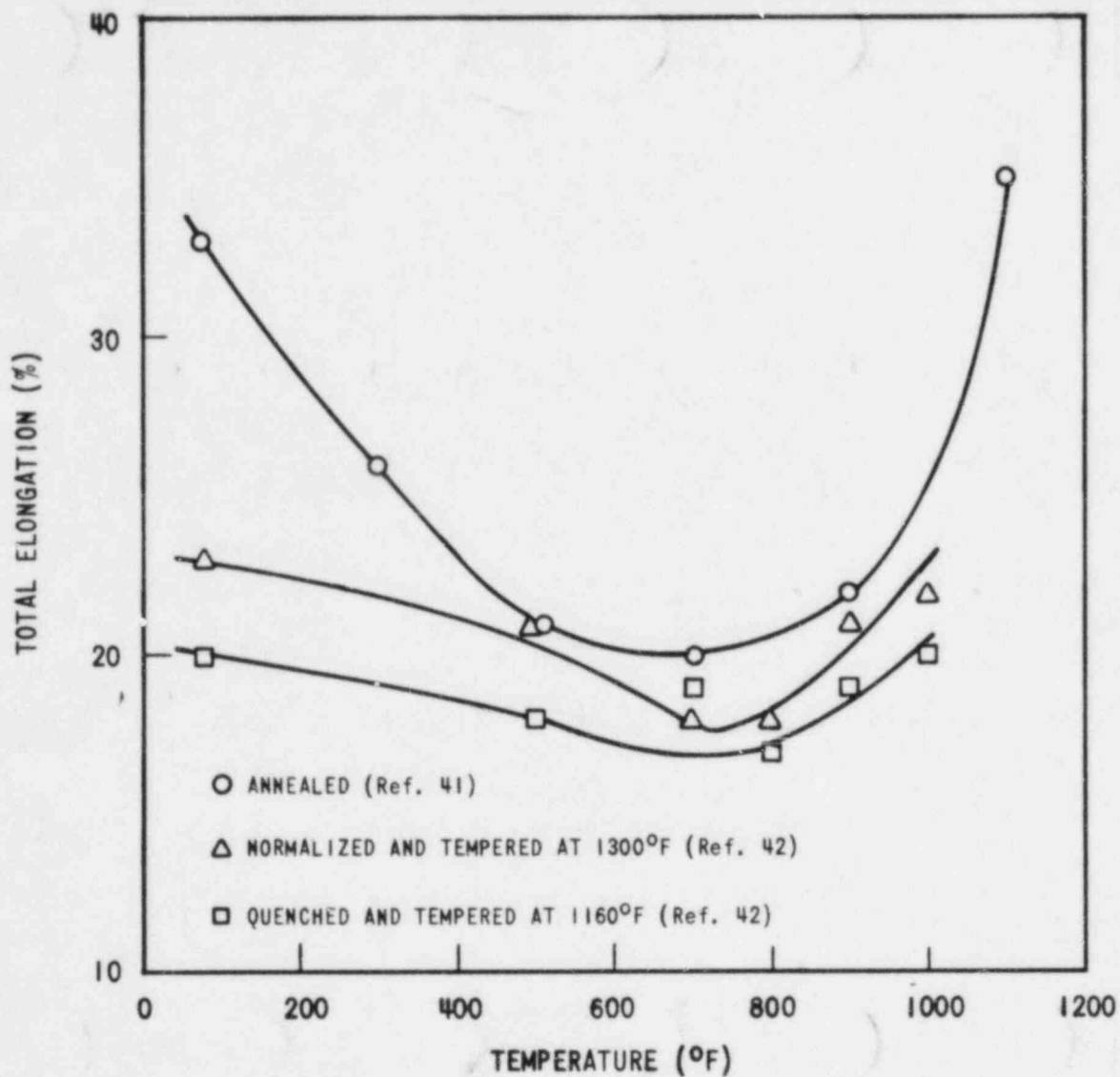


Figure 22. Effect of Heat Treatment on the Total Elongation of 2 1/4 Cr-1 Mo Steel

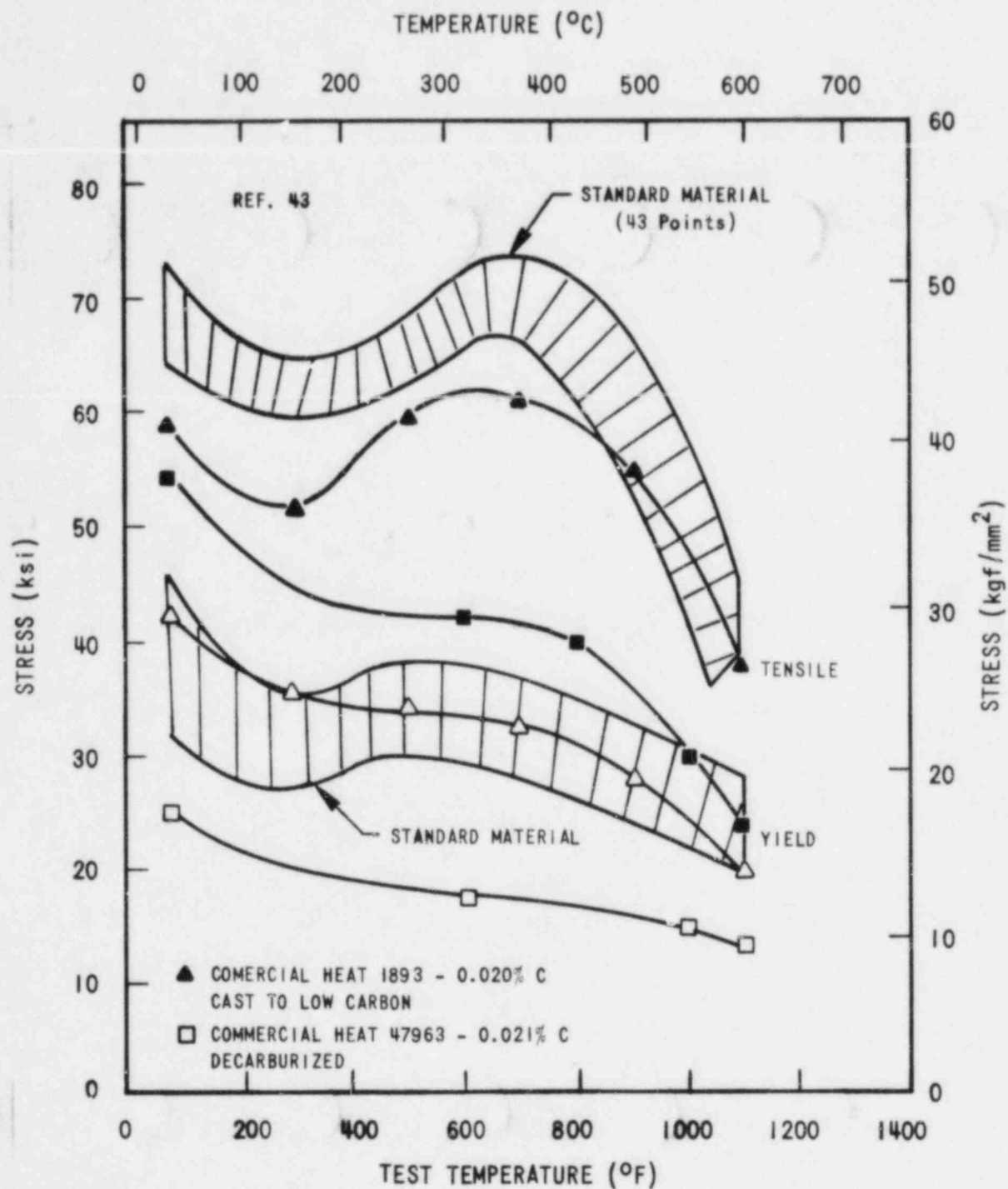


Figure 23. Yield and Tensile Strengths of Low-Carbon 2 1/4 Cr-1 Mo Steels

6129-23



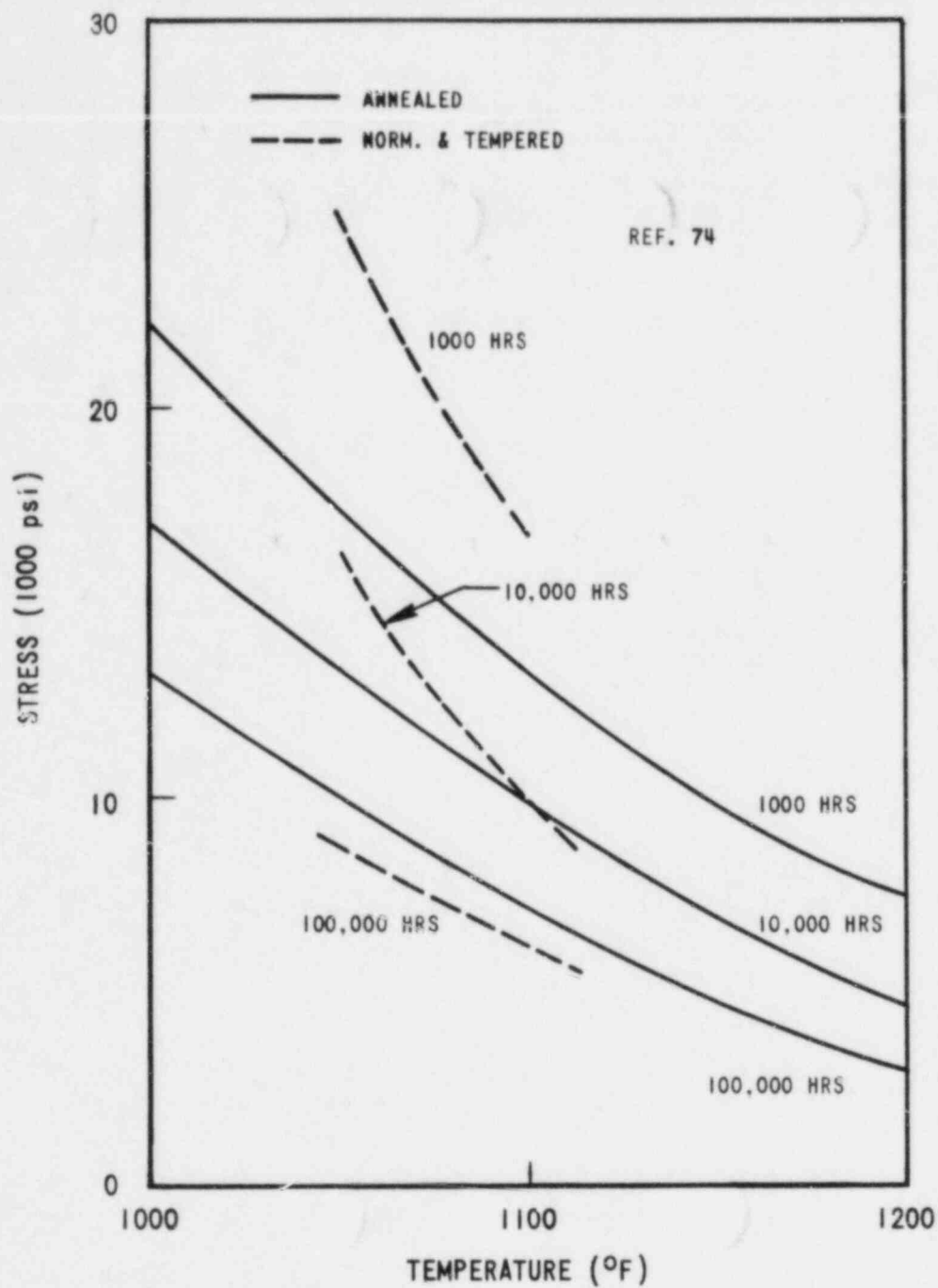


Figure 24. Summary of the Creep Rupture Properties of Annealed and Normalized and Tempered 2.25 Cr-1 Mo Material

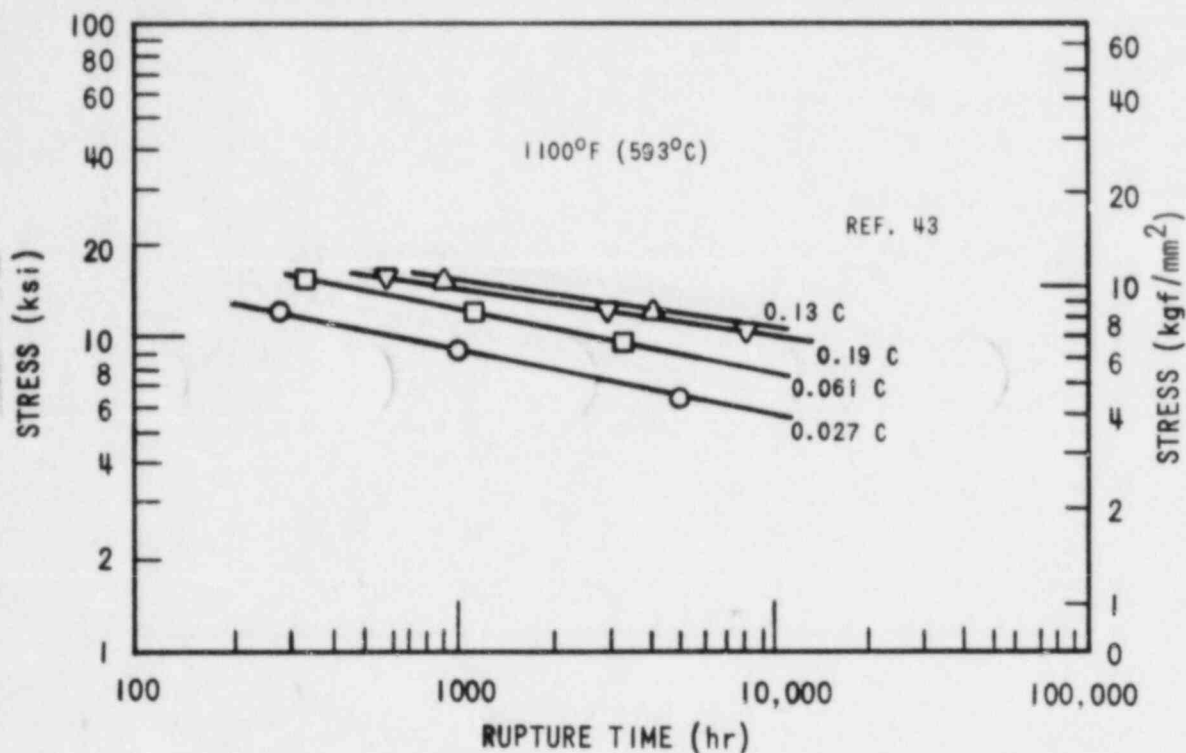


Figure 25. Stress-Rupture Curves for Laboratory Heats of 2 1/4 Cr-1 Mo Steels with Different Carbon Contents

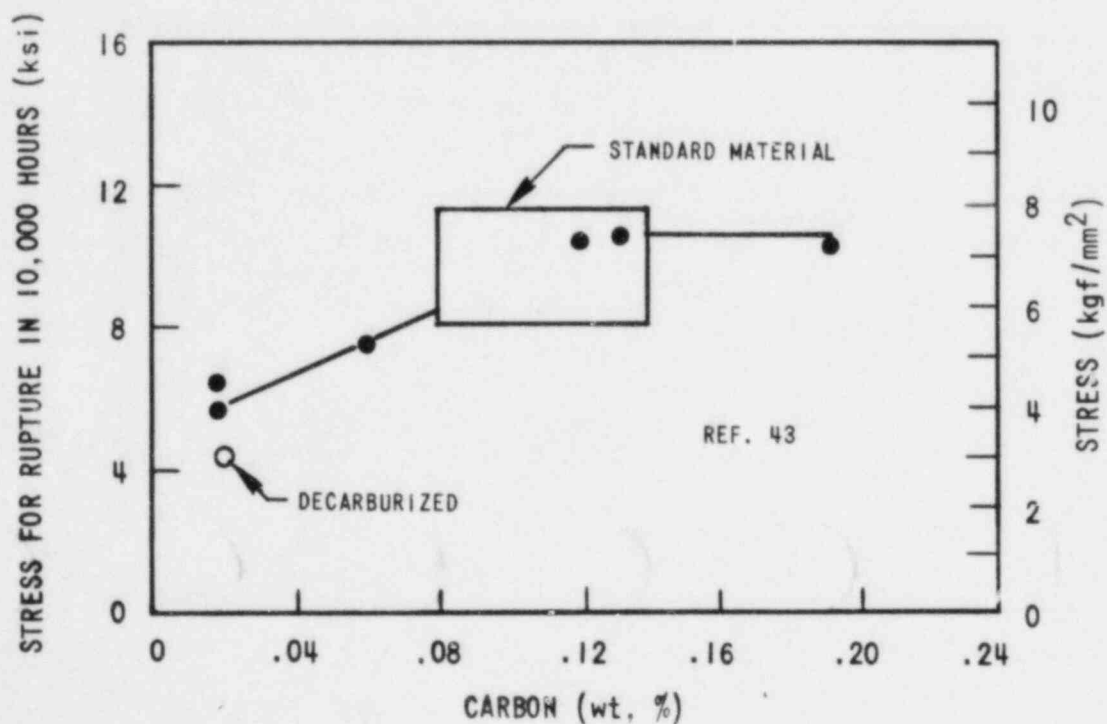
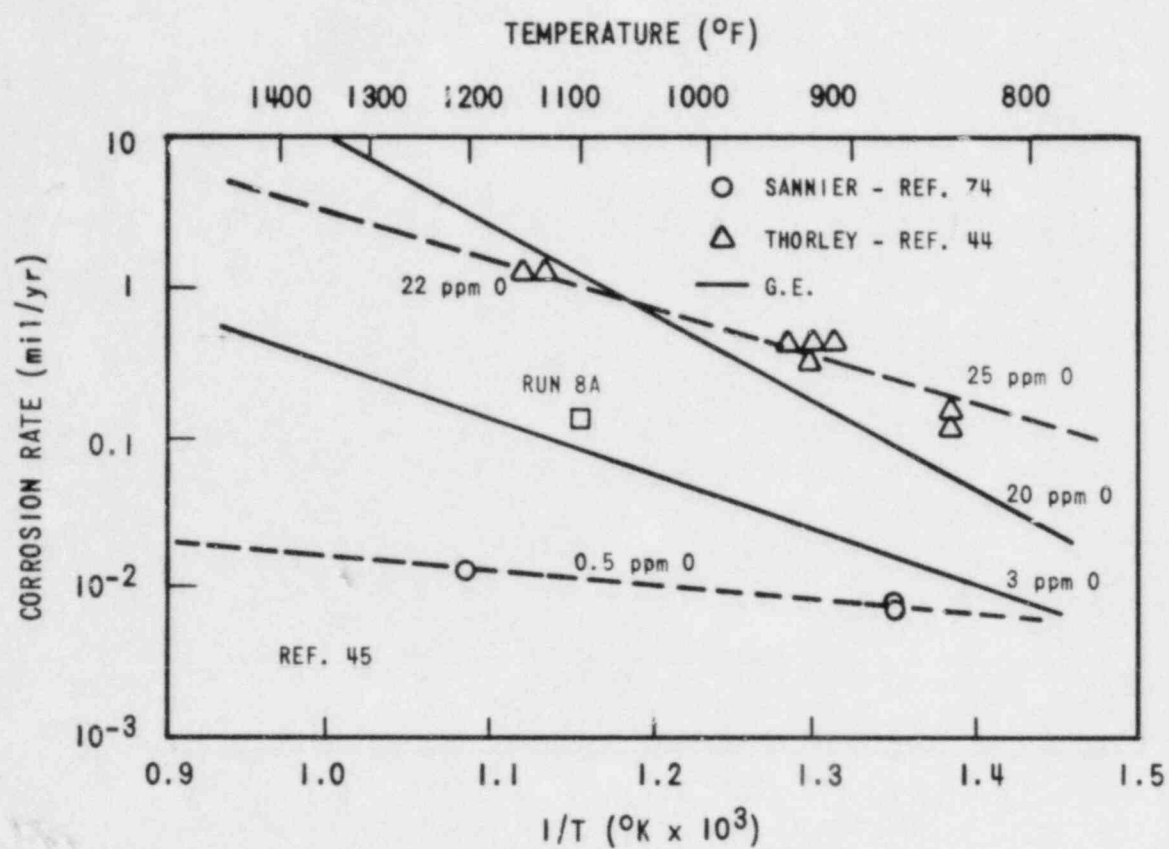


Figure 26. Effect of Carbon Content on the Rupture Strength of 2 1/4 Cr-1 Mo Steel at 1100°F (593 C)



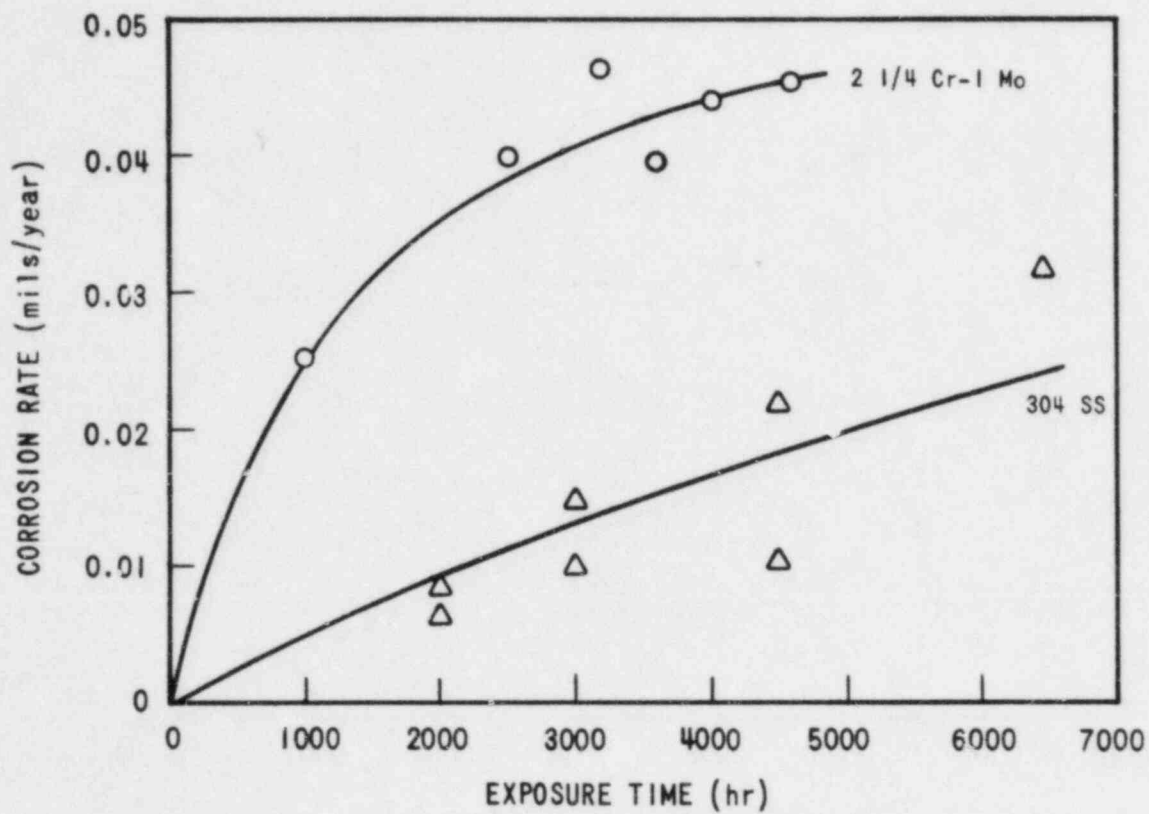


Figure 28. Corrosion Rates for 2 1/4 Cr-1 Mo and Type 304 Stainless Steel in Liquid Sodium at 1200°F

6129-28

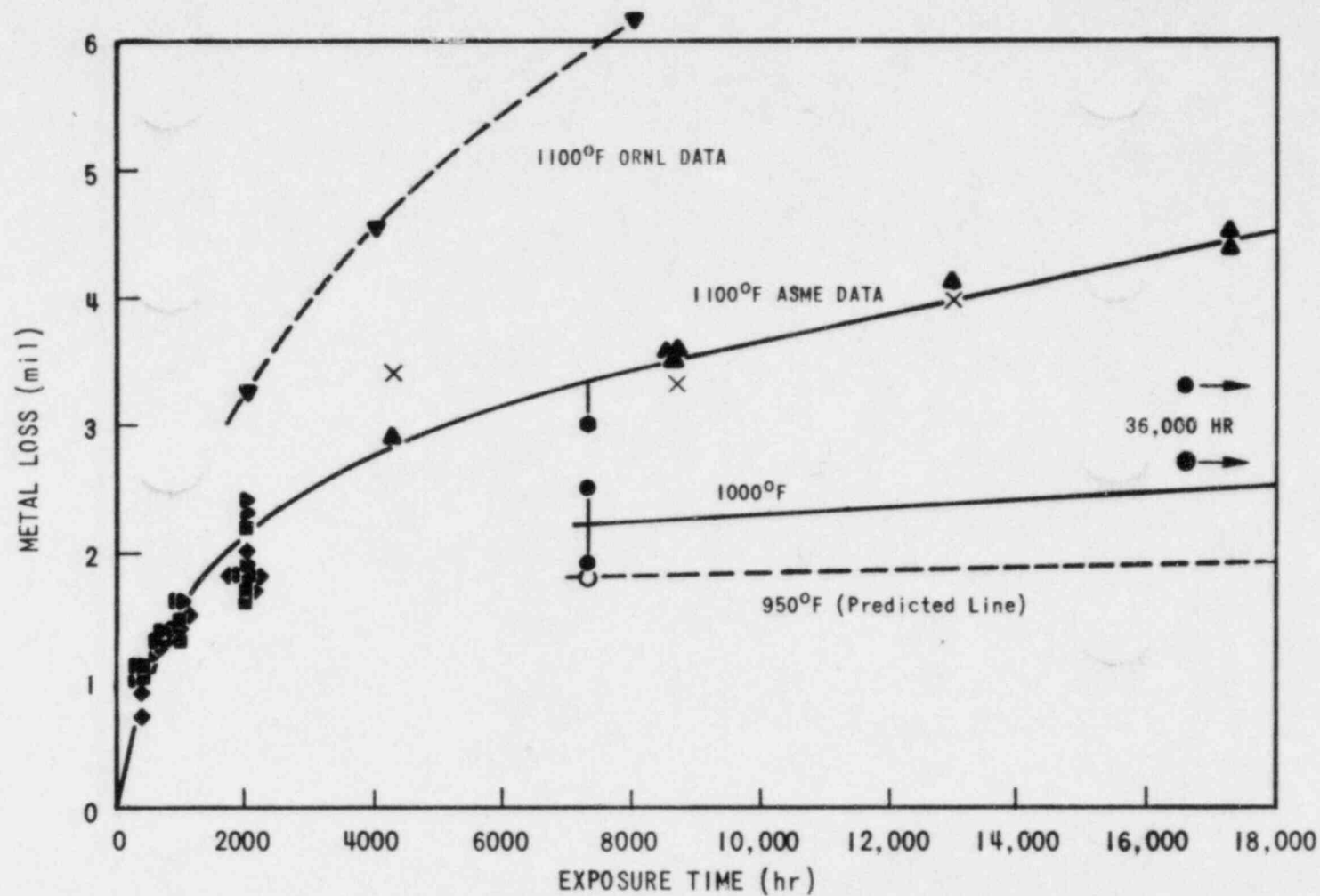


Figure 29. Corrosion of 2 1/4 Cr-1 Mo in Superheated Steam at 950, 1000 and 1100°F

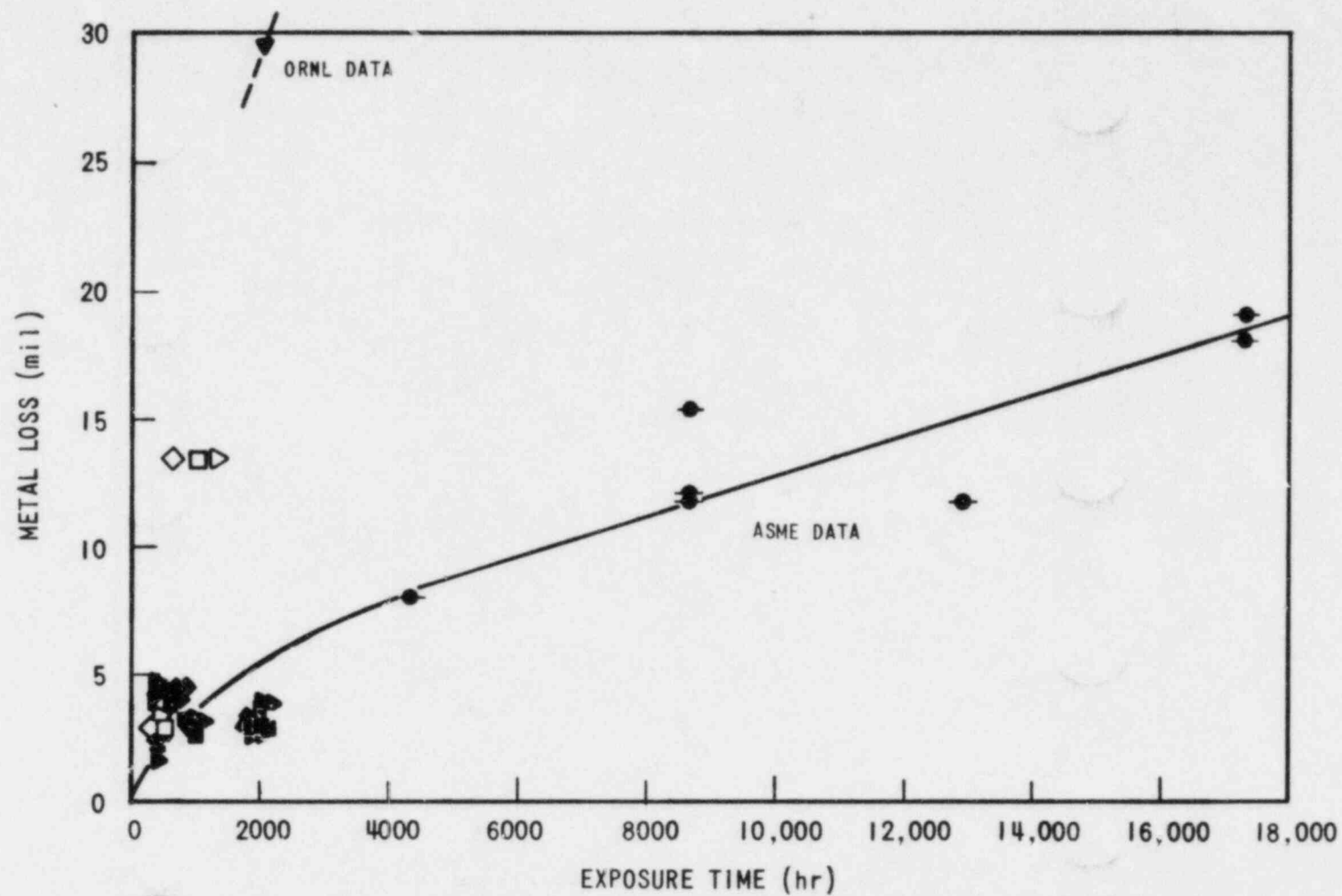
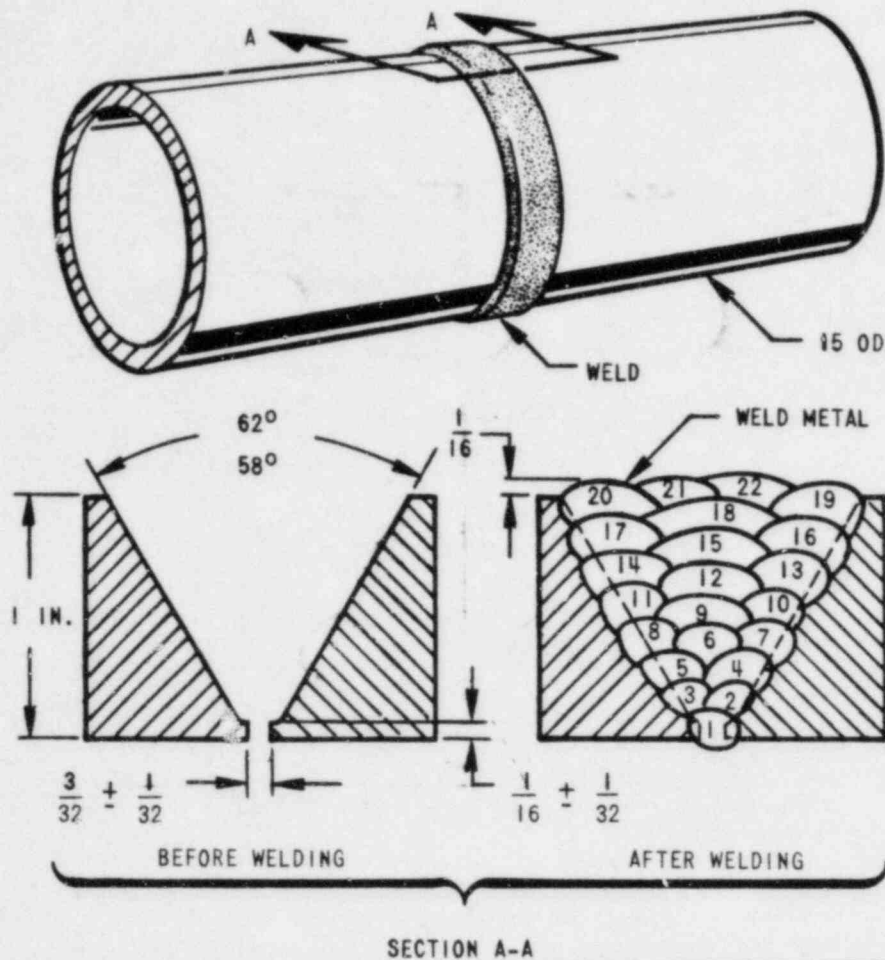


Figure 30. Corrosion of 2 1/4 Cr-1 Mo in Superheated Steam at 1200°F

TYPE 304 STAINLESS STEEL WELDED TO 2.25 Cr, 1 Mo  
STEEL; NICKEL ALLOY FILLER METAL (ENiCrFe-3)



#### CONDITIONS FOR SHIELDED METAL-ARC WELDING

JOINT TYPE.....	BUTT
WELD TYPE.....	SINGLE-U GROOVE
WELDING POSITION.....	HORIZONTAL ROLLED
NUMBER OF PASSES.....	22
PREHEAT AND INTERPASS TEMPERATURE.....	500°F (a)
POSTHEAT.....	1350°F (b)
ELECTRODE WIRE.....	1/8 AND 5/32-IN.-DIA ENiCrFe-3(c)
POWER SUPPLY.....	300-AMP MOTOR-GENERATOR
CURRENT (dcrp) AND VOLTAGE:	
TACK WELDING, AND PASS 1.....	60 AMP, 21 V
PASSES 2 AND 3.....	90 AMP, 23 V
PASSES 4 TO 22.....	120 AMP, 24 V

(a) ONLY THE FERRITIC LOW-ALLOY STEEL SIDE. HEATING WAS BY INDUCTION. (b) HEATED AT 1350°F FOR 1 HOUR BY INDUCTION AND COOLED TO 300°F IN STILL AIR. (c) 1/8-IN.-DIAMETER WIRE FOR PASSES 1 TO 3, AND 5/32-IN.-DIAMETER WIRE FOR PASSES 4 TO 24

Figure 31. Assembly of a Low-Alloy Steel Pipe and a Stainless Steel Pipe that was Joined with Crack-Free Welds by use of a Nickel Alloy Filler Metal

6129-31



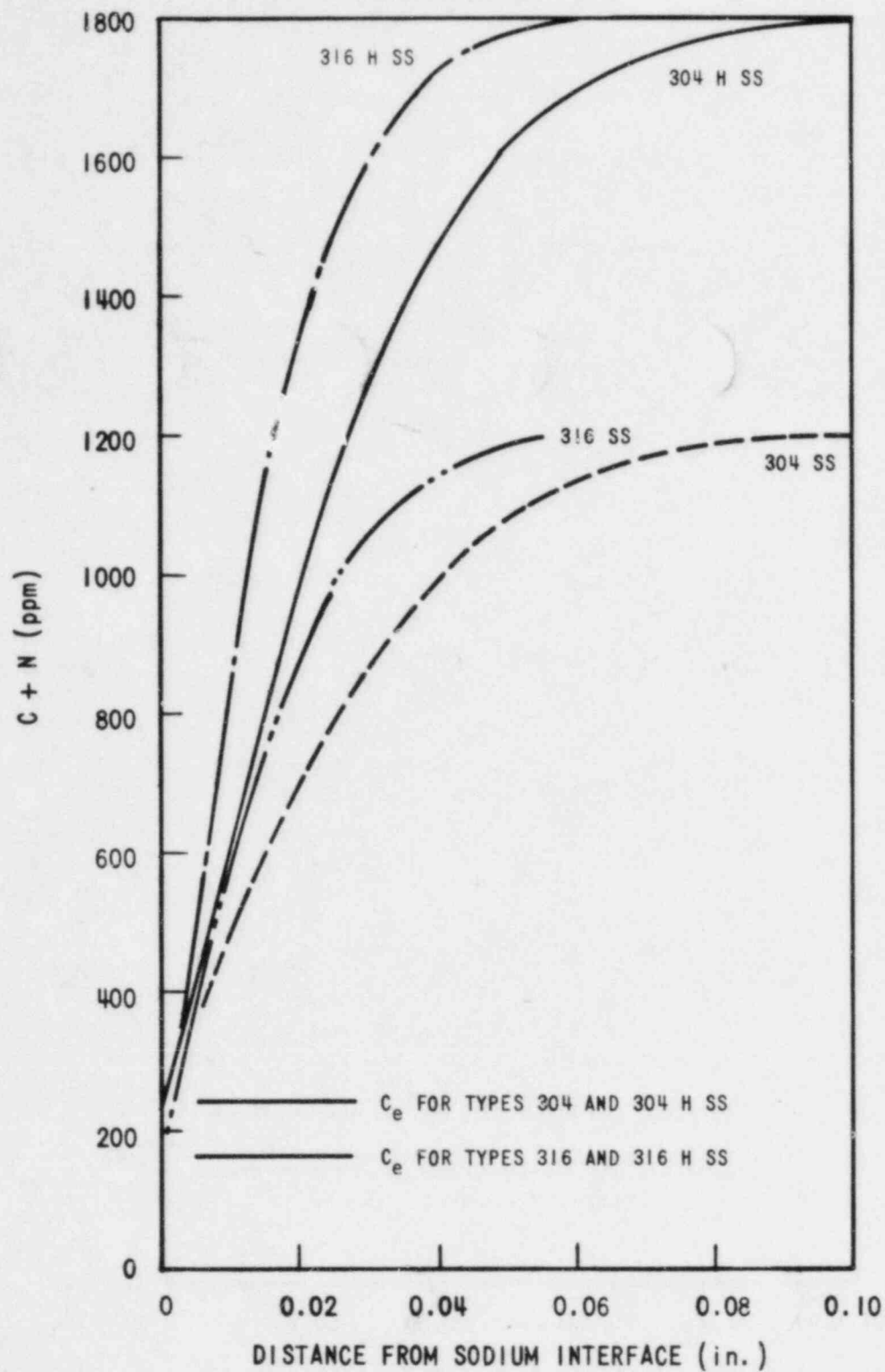


Figure 32. Interstitial Gradients in Primary System Hot Leg Components

6129-32

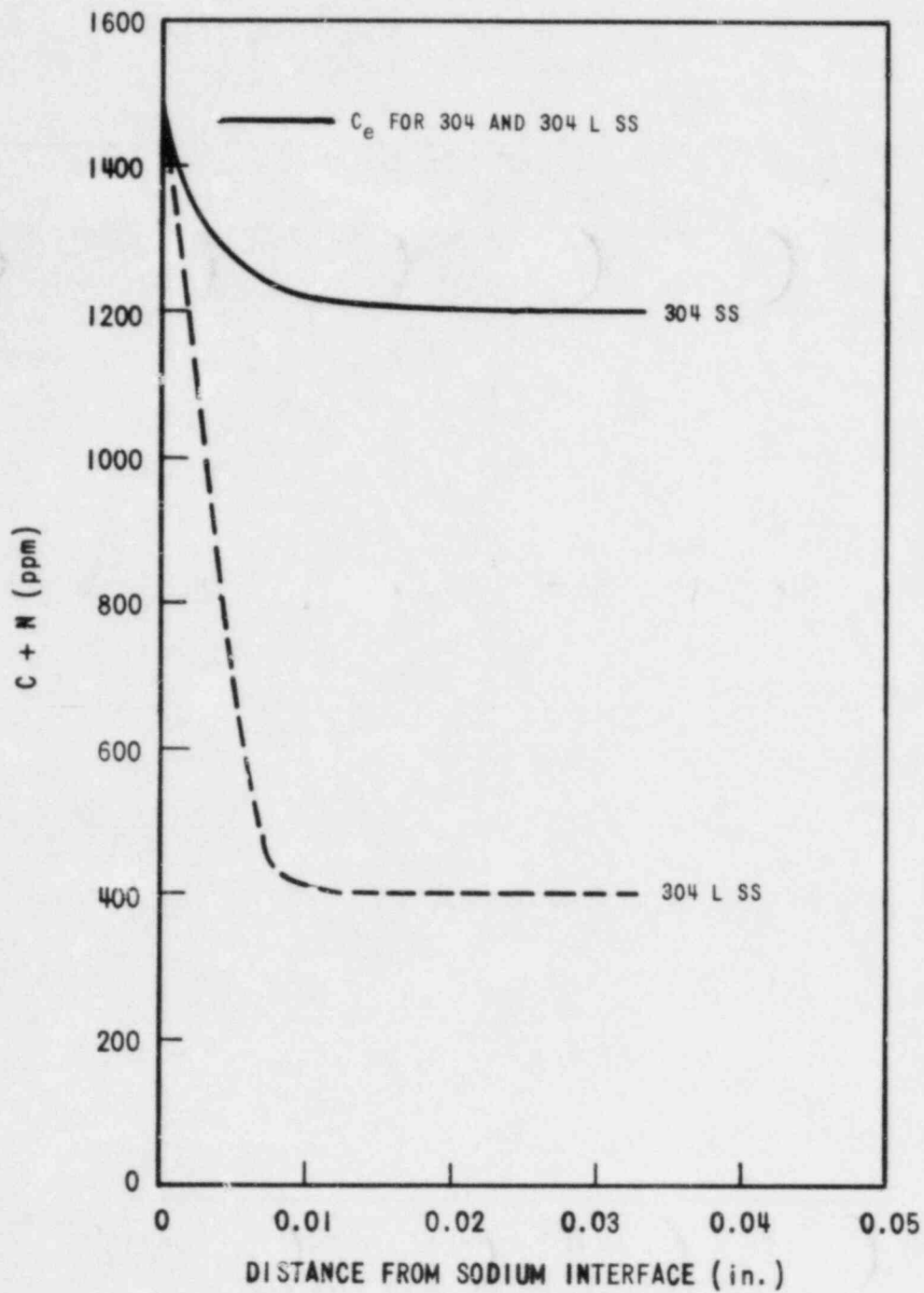


Figure 33. Interstitial Gradients in Primary System Cold Leg Components

6129-33

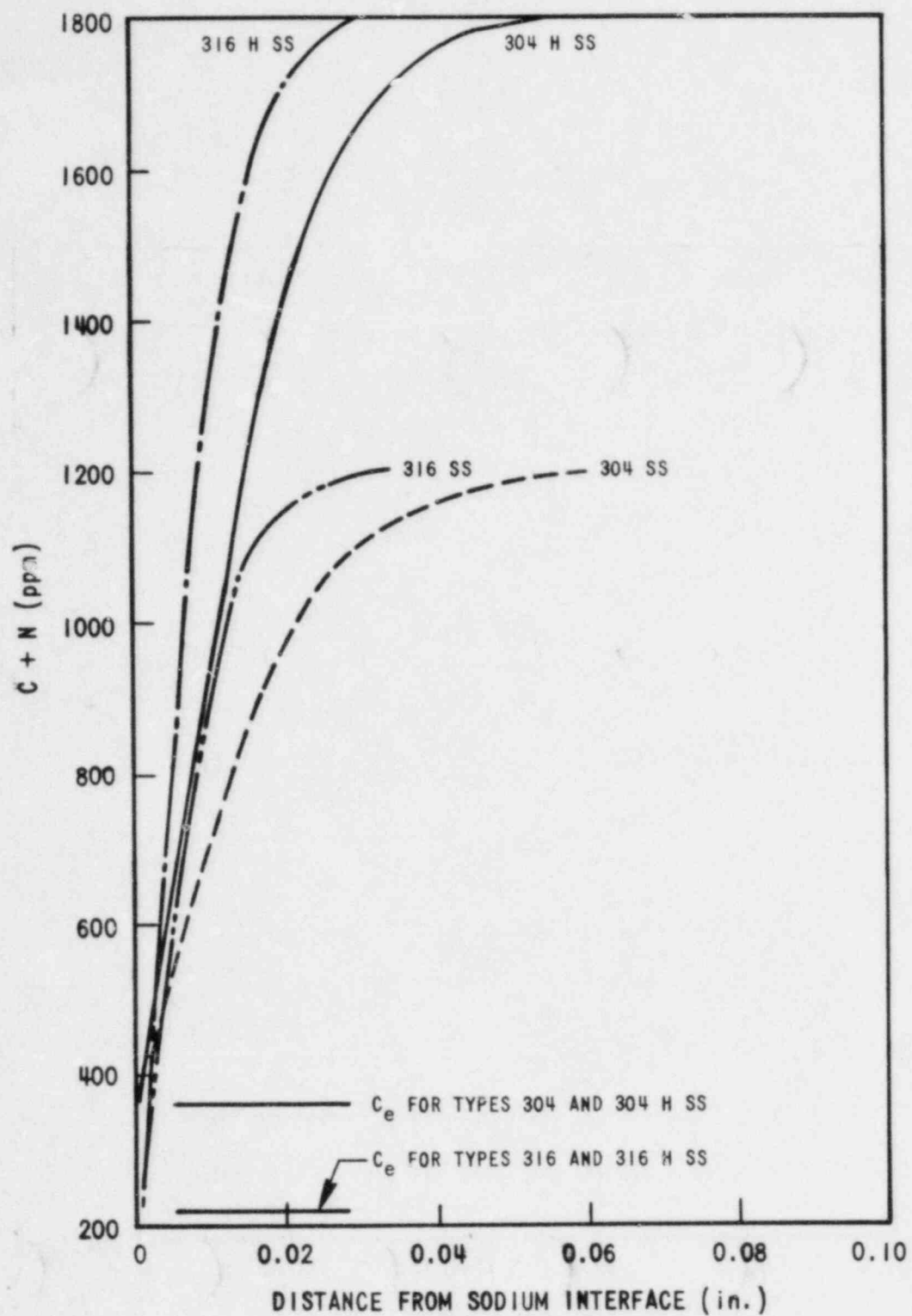


Figure 34. Interstitial Gradients in Intermediate System Hot Leg Components

6129-34

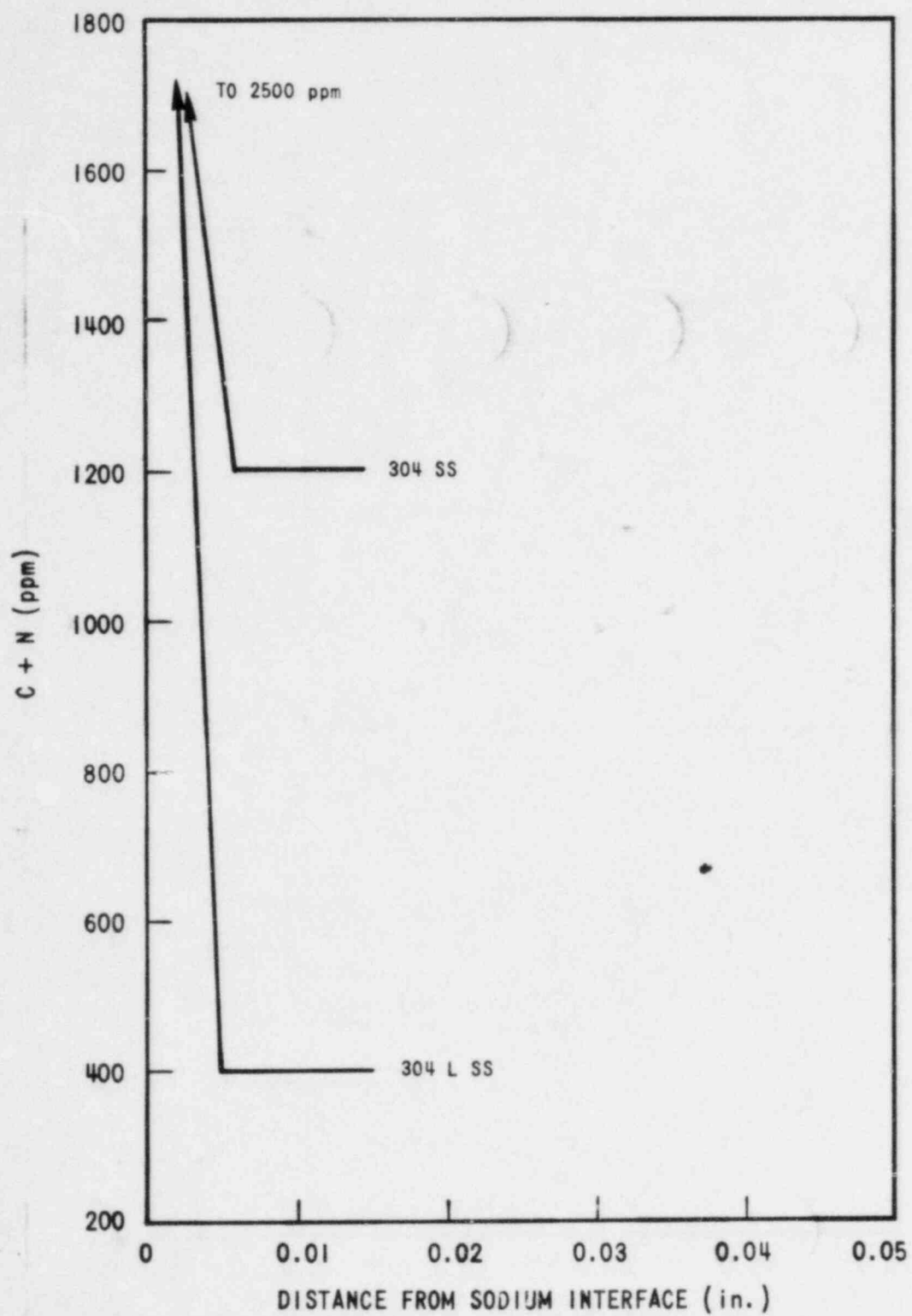


Figure 35. Interstitial Gradients in Intermediate System Cold Leg Components

6129-35