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Optimized Specifications for Types 304 and 316 Stainless Steel in Long-Term High-Temperature Service Applications

C. R. Brinkman
V. K. Sikka
B. L. P. Booker

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OPTIMIZED SPECIFICATIONS FOR TYPES 304 AND 316 STAINLESS STEEL IN LONG-TERM HIGH-TEMPERATURE SERVICE APPLICATIONS

C. R. Brinkman, V. K. Sikka, and B. L. P. Booker

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OPTIMIZED SPECIFICATIONS FOR TYPES 304 AND 316 STAINLESS STEEL IN LONG-TERM HIGH-TEMPERATURE SERVICE APPLICATIONS*

C. R. Brinkman, V. K. Sikka, and B. L. P. Booker

ABSTRACT

The mechanical properties of types 304 and 316 stainless steel vary considerably from heat to heat at temperatures within the creep range. The objective of this effort was to identify the magnitude and causes of these variations and then to recommend improved material purchasing specifications that would reduce data scatter and optimize performance of these steels for high-temperature service. Mechanical properties of interest included tensile (elastic and plastic), creep and creep rupture, strain-controlled fatigue, crack propagation, hardening parameters, and toughness. The importance of controlling such variables as grain size, levels of warm or cold work, carbon and nitrogen content, and residual element content are discussed in terms of the variability in the above properties.

We compare British, French, and current U.S. specifications for austenitic stainless steels to be employed as structural materials in fast breeder reactor service. Suggestions are given for target ranges within current specifications as goals to be sought to decrease data scatter. Recommendations are also given as "good practice" to be followed when practical in procuring these steels to minimize data scatter, adjust ferrite content, or improve a specific mechanical property such as creep ductility, creep strength, time-dependent fatigue, or toughness.

INTRODUCTION

Certain mechanical properties of types 304 and 316 stainless steel vary considerably from heat to heat, particularly at temperatures within the creep range. Our objective is to identify the magnitude and causes of these variations and to recommend improved materials purchasing specifications that would reduce data scatter and optimize performance of these steels for fast breeder reactor service.

*Work performed under DOE/RRT AF 15 40 10 3, OR-1.3, Mechanical Properties Design Data.

Variations in mechanical properties of the austenitic stainless steels are well documented and have been attributed to the following for wrought product forms:

1. chemical composition specifications, which have been made fairly broad in order to minimize cost;
2. mechanical property specifications, which permit some variation in levels of cold work in as-produced (mill-annealed) material;
3. variation in thermomechanical processing history from heat to heat and for various product forms, resulting in grain size variations, cold work, and some preferred orientation in mill-annealed material; and
4. metallurgical changes that occur on prolonged elevated-temperature exposure because these steels are metastable at temperatures within the creep range.

Several examples of typical mechanical property variations are given in Table 1, showing ranges of behavior at both room and elevated temperatures. Note, for example, the extensive variations possible in minimum creep rate for both types 304 and 316 stainless steel at a stress that would produce failure in 10^5 h in an average heat. These variations are so wide that it is virtually impossible to calculate with any real assurance creep strain for a given temperature, time, and loading condition unless the lot constant (measure of relative strength) is known with some assurance for the particular heat in question. Wide variations in other properties, such as high-cycle fatigue life and toughness, are also apparent. On the other hand, elastic properties, such as Young's modulus and Poisson's ratio, show little variation for wrought material in the reannealed condition but can show extensive differences in castings or as-deposited weld metal.

Recently, McAfee and Sartory[1] listed ten mechanical and physical properties commonly used in inelastic and failure analysis as follows:

1. bilinearized yield stress, (a) monotonic $\kappa_0(T)$ (b) tenth cycle $\kappa_1(T)$;
2. slope of plastic portion of stress-strain curve $E^p(T)$;

Table 1. Range of measured or calculated representative mechanical property variations in unaged types 304 and 316 stainless steel

	Type 304 stainless steel				Type 316 stainless steel			
	Minimum	Mean	Maximum	Comment	Minimum	Mean	Maximum	Comment
Creep, 538°C								
Rupture life, h	1.2×10^4	1×10^5	1.1×10^6	*a*	2.1×10^4	1×10^5	4.5×10^5	*b*
Minimum creep rate, %/h	5.1×10^{-7}	1.1×10^{-5}	2.2×10^{-4}	*c*	2.3×10^{-6}	2.2×10^{-5}	2.1×10^{-4}	*d*
Strain to onset of tertiary creep, %	0.64	0.98	2.3	Calculated	1.02	1.74	3.02	Calculated
Total creep elongation in 10^5 h, %	5		25	Estimated	5		25	Estimated
Tensile, 25°C, reannealed								
Yield, MPa	170	200	225					
Ultimate strength, MPa	510	580	635					
Young's modulus, GPa	200.1	201.1	201.8	Dynamic	196.7	197.4	198.0	Dynamic
Poisson's ratio	0.282	0.286	0.291		0.290	0.293	0.297	
Tensile, 25°C, mill annealed								
Yield, MPa	202	245	290		200	248	295	
Ultimate strength, MPa	550	600	650		515	570	630	
Young's modulus, GPa	200.0	200.8	201.1	Dynamic				
Poisson's ratio	0.284	0.288	0.290					
Continuous low-cycle fatigue,[e] 593°C, strain rate $4 \times 10^{-3} s^{-1}$, $\Delta\varepsilon_t$ = 1.0%, cycles to failure N_f	2,000	3,200	7,000	Annealed, aged	2,040	3,225	4,000	
Continuous high-cycle fatigue, 593°C, $\Delta\varepsilon_t$ = 0.28%, cycles to failure N_f		550,000		Estimated	2,833,816	21,192,555	68,310,000	*f*
Creep fatigue, strain control, tension hold, cycles to failure N_f	338	624	767	5 Heats 593°C $\Delta\varepsilon_t \approx 1.0\%$	160	560	1,227	2 Heats 550°C $\Delta\varepsilon_t$ = 1.5% 0.5-h hold[g]

Table 1. (Continued)

	Type 304 stainless steel				Type 316 stainless steel			
	Minimum	Mean	Maximum	Comment	Minimum	Mean	Maximum	Comment
Cyclic fatigue crack propagation[h]								
Fracture toughness, 25°C								
Initiation J_c, kJ/m^2	456	1,045	1,635	*i*	369	398	427	1 Heat
Tearing modulus T	308	450	592	*i*		363		1 Heat
Fracture toughness, 427–538°C								
Initiation J_c, kJ/m^2	178	480	781	*i*	212	232	252	1 Heat
Tearing modulus T	378	527	676	*i*		27		1 Heat

[a]Values calculated for a stress level of 167 MPa (24.2 ksi), lot constant C_h = 25.458 ± 0.983.

[b]Values calculated for a stress level of 221 MPa (32.0 ksi), lot constant C_h = –11.870 ± 0.663.

[c]Values calculated for a stress level of 167 MPa (24.2 ksi), lot constant C_h = –33.89 ± 1.319.

[d]Values calculated for a stress level of 221 MPa (32.0 ksi), lot constant C_h = 11.314 ± 0.979.

[e]Strain-controlled fatigue test results.

[f]Includes large- and small-grain material (i.e., ASTM 3.2 and 5.0).

[g]Results strongly dependent on thermal history of material. Heat treatment varied to give indicated range of results.

[h]The effects of metallurgical variables and heat-to-heat variations tend to be minimal under conditions leading to transgranular crack propagation. However, under conditions leading to intergranular or accelerated creep crack propagation, metallurgical variables are likely to be more important.

[i]Values measured in temperature range from 427 to 538°C and are generally thought to be independent of temperature in this range.

3. creep strain versus time at constant stress and temperature $\varepsilon(\sigma, t, T)$;
4. time to rupture $t(\sigma, T)$;
5. fatigue strain range $\varepsilon_t(T, N_f)$;
6. Young's modulus $E(T)$;
7. Poisson's ratio $\nu(T)$;
8. instantaneous thermal expansion coefficient $\alpha(T)$;
9. volumetric heat capacity $C_v(T)$; and
10. thermal conductivity $\kappa(T)$.

To the above list, one must also add for consideration certain other material characteristics and properties such as weldability, toughness, parameters related to crack propagation rates, creep ductility, corrosion resistance during shipment and storage, and, in the case of vessel or in-reactor components, any unusual susceptibility to irradiation-induced embrittlement brought about by the presence of residual elements such as boron.[2] Then, for analysis by the American Society of Mechanical Engineers (ASME) Code Case N-47 rules, one requires a knowledge of the mechanical properties indicated as 3 through 7 in the above list as well as the ultimate and yield strengths and creep-fatigue properties. Long-term creep-rupture ductility is important because it is indicative of creep-fatigue response in the low-cycle fatigue regime. Hence, it is desirable to minimize scatter in all these properties.

Physical properties indicated in the above list, particularly thermal expansion and heat capacity, are not expected to change very much within the current or any revised chemical specifications for a given steel. These properties tend to be independent of interstitial element content. Thermal conductivity may vary somewhat, but again the variations are not expected to be significant. The thermal conductivities of types 304 and 304L stainless steel are virtually the same.[3] However, significant variations in substitutional elements such as Ni, Cr, and Mo would influence physical property values. In a recent compilation of physical properties (i.e., thermal conductivity, thermal diffusivity, specific heat, instantaneous coefficient of thermal linear expansion, mean coefficient of thermal linear expansion, and thermal linear expansion), Moen made no distinction between the properties of 18% Cr—8% Ni and 18% Cr—8% Ni—N (types 303, 303Se, 304, 304L, 304H, and 304N) grades of stainless steel.[4] Similarly, no differences were shown for the 16% Cr—12% Ni—2% Mo and 16% Cr—12% Ni—2% Mo—N (types 316, cast 316, 316L, 316H, and 316N) grades either, although differences between types 304 and 316 stainless are apparent. The author does state that the estimated uncertainties of ±10% are typical, particularly of elevated-temperature values. Small amounts of cold work are not expected to influence physical properties unless an order-disorder reaction or some other deformation-induced phase change occurs. Therefore, variations in behavior of concern in wrought products of a given steel are those mechanical properties involving primarily both large and small amounts of plastic deformation.

As indicated above, one of the major sources of heat-to-heat variation or data scatter is the current American Society of Testing and Materials (ASTM), ASME, and Nuclear Regulatory Commission standards for types 304 and 316 stainless steel, which allow considerable latitude in the chemical composition, melting practice, and thermomechanical processing history of these steels. Typical ranges of permissible chemical compositions are given in Tables 2 through 4, and Table 5 gives additional Nuclear Regulatory Commission standards requirements. Tables 2 and 3 also give the specific compositions of the ORNL reference heats of types 304 and 316 stainless steel, respectively, for comparison. Also given are chemical specifications imposed by a number of foreign countries or projects on these or very similar steels intended for use as structural materials in LMFBR service. Finally, Table 5 lists the chemical requirements imposed by the indicated Nuclear Regulatory Commission specifications for several product forms (i.e., plate and bar intended for use as core component material). Note the restricted range of carbon and the limitations on nitrogen and boron content.

It should be noted that, in the United States, nuclear grade stainless steels (types 304NG and 316NG) have been developed for use in the water-cooled reactor industry to avoid sensitization and stress corrosion cracking. Nuclear grade stainless steel contains no more than 0.02% C with nitrogen levels of about 0.07% to maintain strength and the austenitic structure.[5]

Table 2. National nuclear specifications for type 304 stainless steel

Element	United States: ASME general[a]	United States: ORNL reference heat ASTM A 240 51-mm plate	Germany, SNR-300 DIN 1-4948	Japan, Industrial Standards, Monju[b]: Forgings, pressure vessels JIS G3214	Japan, Industrial Standards, Monju[b]: Plates and sheets JIS G4304
Content, wt %					
Carbon	0.04–0.08	0.047	0.04–0.08	0.04–0.08	0.04–0.08
Nitrogen		0.031			
Silicon	1.00 Max	0.10	<0.75	1.0 Max	1.0 Max
Manganese	2.00 Max	1.22	<2.0	2.0 Max	2.0 Max
Nickel	8.00–10.50	9.58	10.0–12.0	8.0–11.0	8.0–10.50
Chromium	18.00–20.00	18.5	17.0–19.0	18.00–20.00	18.00–20.00
Molybdenum		0.10	<0.5		
Sulfur	0.03 Max	0.012	<0.02	0.030 Max	0.030 Max
Phosphorus	0.045 Max	0.029	<0.035	0.040 Max	0.040 Max
Niobium		0.008			
Cobalt		0.05		<0.25	0.25
Titanium		0.003			
Vanadium		0.037			
Tungsten		0.022			
Copper		0.10			
Tin		0.02			
Heat treatment, temperature, °C	1038–1121[c]	1093, Rapid cool	1020–1100, Water or air quench	1020–1035, Water quench	1100, Water quench

[a]ASME Code Case N-47 requires a minimum specified carbon content of 0.04%.

[b]Japanese restrict δ-ferrite in "mother metal" to values of less than 1% and are also attempting to limit grain size to a maximum of ASTM 0.

[c]*Source*: "Heat Treating of Stainless Steel and Heat-Resisting Alloys," *Metals Handbook*, 8th ed., vol. 2, American Society for Metals, Metals Park, Ohio, 1971. Note that ASME Code specifies a minimum room-temperature yield strength of 207 MPa (30 ksi).

Table 3. United States nuclear specifications for type 316 stainless steel

Element	ASME general[a]	Plate M5-19T core components[b]	Bar M7-23T core components[c]	ORNL reference heat	
				ORNL MC-316	Ladle analysis
Content, wt %					
Carbon	0.04–0.08	0.04–0.06	0.04–0.06	0.05–0.08	0.057
Nitrogen		0.01 Max	0.01 Max	0.05 Max	0.03
Boron		0.001 Max	0.002 Max	0.003 Max	0.0005
Silicon	1.0 Max	0.50–0.75	0.50–0.75	0.02–0.6	0.58
Manganese	2.0 Max	1.5–2.0	1.0–2.0	1.5–2.0	1.86
Nickel	10.0–14.0	13.0–14.0	13.0–14.0	13.25–13.75	13.48
Chromium	16.0–18.0	17.0–18.0	17.0–18.0	16.5–17.5	17.25
Molybdenum	2.0–3.0	2.0–3.0	2.0–3.0	2.0–2.5	2.34
Sulfur	0.03 Max	0.010 Max	0.01 Max	0.01–0.025	0.019
Phosphorus	0.045 Max	0.020 Max	0.04 Max	0.02–0.03	0.024
Niobium		0.050 Max	0.05 Max	0.02 Max[d]	<0.1
Cobalt		0.050 Max	0.05 Max	0.10 Max	0.02
Titanium				0.02 Max	0.02
Vanadium		0.20 Max	0.05 Max		
Copper		0.04 Max	0.04 Max	0.20 Max	0.10
Aluminum		0.05 Max	0.05 Max		
Arsenic		0.03 Max	0.03 Max		
Tin					0.004
Heat treatment, temperature, °C	1038–1121,[e] Water quench or rapid cool	1038–1121,[e] Water quench or rapid cool	1038–1121,[e] Water quench or rapid cool	1065 ± 30, Air cool	

[a]ASME Code Case N-47 requires a minimum specified carbon content of 0.04%.

[b]Grain size for 50.8- to 254-mm-diam (2–10-in.-) bar shall be ASTM 3 or finer, February 1982 requirement.

[c]Grain size for 50.8-mm (2-in.) or thicker plate shall be ASTM 3 or finer, March 1973 requirement.

[d]Nb + Ta < 0.20.

[e]*Source*: "Heat Treating of Stainless Steel and Heat-Resisting Alloys," *Metals Handbook*, 8th ed., vol. 2, American Society for Metals, Metals Park, Ohio, 1971. Note that the ASME Code specifies a minimum room-temperature yield strength of 207 MPa (30 ksi). The Nuclear Regulatory Commission standards specify a minimum solution heat treatment of 1038°C.

Table 4. Foreign national chemical specifications for type 316 stainless steel

Element	Germany, DIN 1-4919	France		United Kingdom Atomic Energy Authority		Italy	Japan, Industrial Standards,[a] Monju	
		Water reactor CN	Super Phenix SPH	Present CDFR (ISO)	Revised CDFR/CFRX (proposed)		Forgings, Pressure vessels JIS G3214	Plates and sheets JIS G4304
Content wt %								
Carbon	0.04–0.08	<0.045	0.03 Max	0.04–0.09	0.03–0.06	0.08 Max	0.04–0.08	0.04–0.08
Nitrogen		<0.08	0.06–0.08		0.04–0.07			
Boron		<0.0035	0.0015–0.0035		0.002–0.005			
Silicon	<0.75	<1.0	0.50 Max	0.75 Max	0.2–0.60	1.5 Max	1.0 Max	1.0 Max
Manganese	<2.0	<2.0	1.6–2.0	1.0–2.0	1.6–2.0	2.0 Max	2.0 Max	2.0 Max
Nickel	12.0–14.0	11.5–12.5	12.0–12.5	12.0–14.0	11.5–12.5	1.0–14.0	10.0–14.0	10.0–14.0
Chromium	16.0–18.0	17.0–18.2	17.0–18.0	16.0–18.0	16.5–18.0	16.0–18.0	18.00–20.0	18.0–20.0
Molybdenum	2.0–2.5	2.3–2.8	2.3–2.7	2.0–2.75	2.0–2.75	2.0–3.0	2.00–3.00	2.0–3.0
Sulfur	<0.030	<0.030	0.025 Max	0.03 Max	0.015 Max	0.03 Max	0.03 Max	0.03 Max
Phosphorus	<0.034	<0.035	0.035 Max	0.045 Max	0.030 Max	0.045 Max	0.04 Max	0.04 Max
Niobium					0.02 Max			
Cobalt		<0.20			0.15 Max		<0.25	<0.25
Titanium					0.02 Max			
Zirconium					0.02 Max			
Vanadium					0.01 Max			
Tungsten					0.01 Max			
Copper		0.40			0.25 Max			
Aluminum					0.05 Max			
Arsenic, Antimony, Bismuth, Tin					0.03 Max total			
Heat treatment, temperature, °C	1020–1100, Water or air quench	1050–1100, Water quench	1050–1100, Water quench	1020–1070, Rapid cool	1050–1100, Rapid cool	1030, Quench	1020–1035, Rapid cool	1100, Rapid cool

[a]Japanese restrict δ-ferrite in "mother metal" to values of less than 1% and are also attempting to limit grain size to a maximum of ASTM 0.

Table 5. Nuclear Regulatory Commission Standards relating to ASME Code Case N-47 with additional mechanical property-related requirements

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Plate

NE M 5-23T *Steel Plates for Nuclear and Other Special Applications (ASME SA-647 with Additional Requirements)* Covers 304, 316, and 2 1/4 Cr-1 Mo or (Grade 22 SA-387)

Additional Requirements:

Austenitic steels shall be heated to a minimum of 1900°F (1038°C) and then shall be quenched in water or rapidly cooled by other means. Rapid cooling by other means is acceptable only when the cooling rate attained is sufficient to prevent reprecipitation of chromium carbides as demonstrated by freedom from intergranular attack.

Austenitic material intended for use at temperatures over 800°F (427°C) in ASME Code applications shall have a minimum carbon content of 0.04%, or greater when specified by the purchaser. In addition, when repair by welding is permitted, the delta ferrite of A8 weld filler metal, except type 16-18-2, shall be 5 to 9.2%; and the delta ferrite of type 16-8-2 shall be 5% maximum.

Bolting

NE M 6-3T *Alloy Steel Bolting Material for High-Temperature Service (ASME SA-193 with Additional Requirements)*

Additional Requirements:

Austenitic materials intended for use at temperatures over 800°F (427°C) in ASME Code applications shall have a minimum carbon content of 0.04% or greater as specified by the purchaser.

Pipe and Tubing

NE M 3-34T *Pipe and Tubing for Nuclear and Other Special Applications (ASME SA-655 with Additional Requirements)*

Additional Requirements:

Austenitic steels shall be heated to a minimum of 1900°F (1038°C) and then shall be quenched in water or rapidly cooled by other means. Rapid cooling by other means is acceptable only when the cooling rate attained is sufficient to prevent reprecipitation of chromium carbides as demonstrated by freedom from intergranular attack.

Table 5. (Continued)

Austenitic material intended for use at temperatures over 800°F (427°C) in ASME Code applications shall have a minimum carbon content of 0.04%, or greater when specified by the purchaser. In addition, when repair by welding is permitted, the delta ferrite of A8 weld filler metal, except type 16-8-2, shall be 5 to 9.2; and the delta ferrite of type 16-8-2, shall be 5% maximum.

NE M 2-21T *Forgings and Bars for Nuclear and Other Special Applications (ASME SA-654 with Additional Requirements)*

Additional Requirements:

Austenitic steels shall be heated to a minimum of 1900°F (1038°C) and then shall be quenched in water or rapidly cooled by other means. Rapid cooling by other means is acceptable only when the cooling rate attained is sufficient to prevent reprecipitation of chromium carbides as demonstrated by freedom from intergranular attack.

Austenitic material intended for use at temperatures over 800°F (427°C) in ASME Code applications shall have a minimum carbon content of 0.04%, or greater when specified by the purchaser. In addition, when repair by welding is permitted, the delta ferrite of A8 weld filler metal, except type 16-8-2, shall be 5 to 9.2; and the delta ferrite of type 16-8-2, shall be 5% maximum.

The comparisons given in Tables 2 through 4 indicate that the Europeans and the British have evolved or are evolving specifications for type 316 stainless steel. France has a nuclear specification (316L SPH) that is a subclass of its national specification and was developed for Super Phenix. Currently, the United Kingdom is developing a specification for CDFR similar to that of the French. Representatives from both the United Kingdom and France have stated to the author that little or no cost penalty was taken in producing a subclass with a more restrictive chemical composition.

A comparison of the short-term tensile properties of type 316 stainless steel reported by the United Kingdom and the European countries[6] listed in Table 4 and data derived in the United States indicated the following.

1. No significant differences were noted in the average room- and elevated-temperature tensile properties as obtained by French, German, Italian, United Kingdom, and U.S. investigators. These properties were obtained on multiple heats and product forms from about 3000 tests.

2. A minimum in tensile ductility occurs between 300 and 400°C in all material compared, as shown for average properties in Fig. 1.

3. United States investigators noted slight decreases in the average ultimate and yield strengths on reannealing, particularly for material in the form of bars. French investigators conducted tensile tests on six heats of steel after re-solution treatment for 1 h at 1150°C. Such a treatment had little effect on the yield strength of four of the six heats. It did, however, increase the grain size significantly, and any changes noted were attributed to differences in grain size rather than to the level of cold work.

4. The average yield strength of United Kingdom heats and U.S. reannealed material was slightly lower than values obtained from other countries.

5. The lower confidence limits, taken as the average minus twice the standard deviation (2σ), of yield strength, ultimate tensile strength, elongation, and reduction of area are significantly higher for the French steel than for the German, Italian, and United Kingdom data. This was attributed to the smaller spread in the data brought about by the narrow

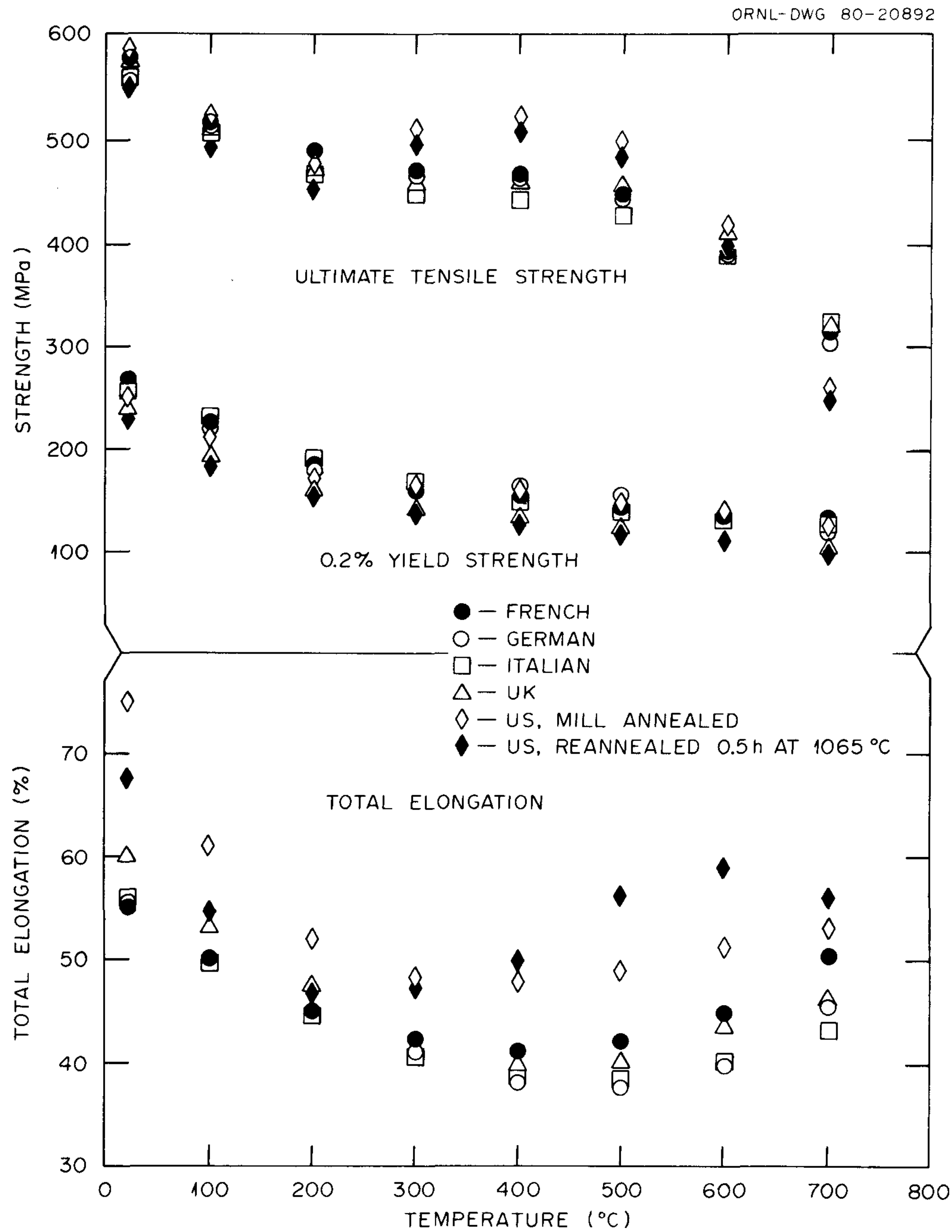

Fig. 1. Comparison of the average tensile properties obtained from many heats of type 316 stainless steel determined in several countries as indicated.

specification of the French steel. Thus, in terms of tensile property variations (heat to heat), the French have demonstrated that differences can be minimized by more restrictive composition limitations. From expressions for yield strength as a function of temperature, the reduction in scatter for French data is apparent.

Country	Number of tests	$1.96\sigma_{ys}$ (MPa)
Italy	670	51.1
France	895	28.4*
Germany	560	57.6
United Kingdom	526	53.8

The French version of type 316 stainless steel (316L SPH) is a 316L grade with reduced C and Si contents and slightly increased N and B concentrations to maintain high-temperature strength. British investigators,[7] in proposing their revised CDFR specification,[8] made similar modifications and concluded that the optimum carbon content is about 0.05 wt %. They also considered weldability and hot working to avoid hot cracking in adjusting these specifications as shown in Fig. 2 in the form of a DeLong representation.[9] Note the more restricted ferrite window or the increased probability of having a small amount of ferrite present for the revised CDFR and Super Phenix materials in comparison with several heats of type 316 stainless steel produced in the United States and under study at ORNL.

In the United States some design and fabricating organizations utilize what is known as "good practice" in making slight adjustments within the Nuclear Regulatory Commission specifications to achieve similar objectives. For example, it has been reported that Foster Wheeler attempts to achieve a 5 to 9% ferrite content while minimizing sulfur and phosphorus to facilitate welding operations.[10] Further, it attempts to restrict certain residual elements as shown below.

*Note that scatter has been reduced by nearly a factor of 2.

Element	Al	Sb	B	Pb	Se	Sn	V	Zn
Maximum content, ppm	500	200	10	30	150	150	500	100

Observations and conclusions concerning the causes of heat-to-heat variability in a number of mechanical properties of types 304 and 316 stainless steel will now be discussed. These properties include tensile, creep rupture and creep, creep-fatigue (time-dependent fatigue), cyclic crack propagation, and toughness.

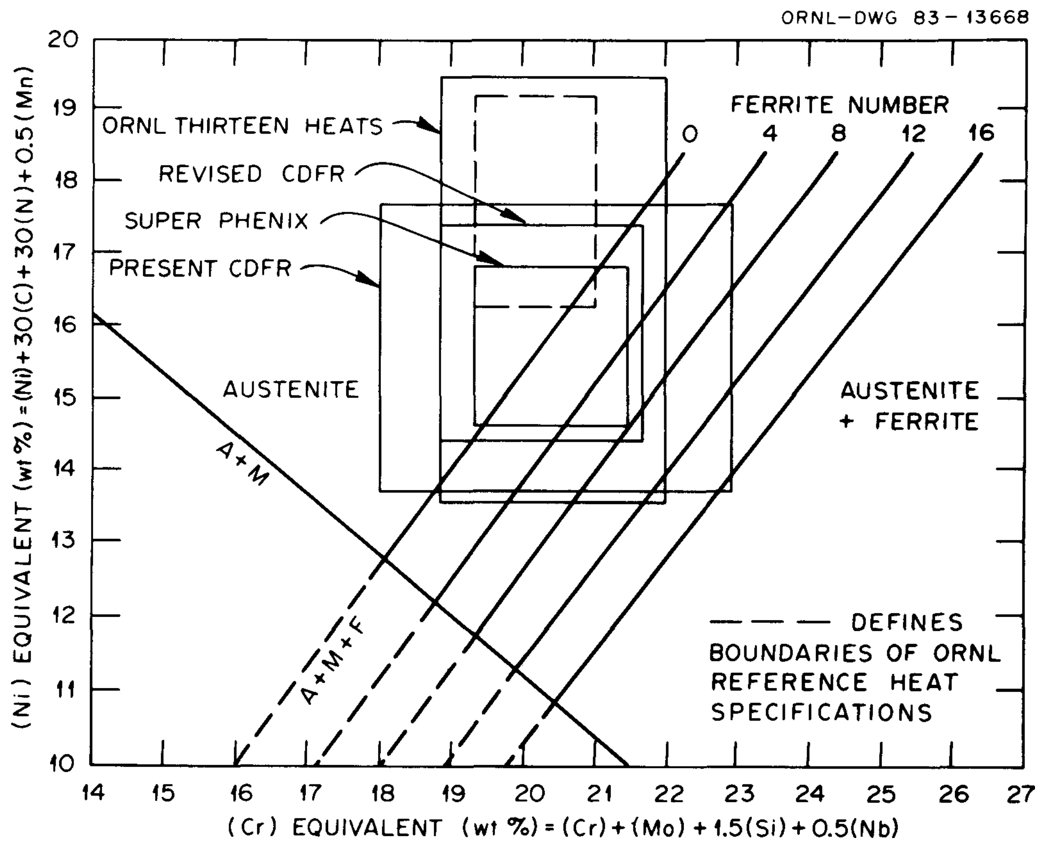

Fig. 2. Comparison of ferrite window chemical specifications for several LMFBR projects with ORNL heats of type 316 stainless steel (DeLong representation).

VARIATIONS IN SPECIFIC MECHANICAL PROPERTIES

TENSILE

The short-term strength properties of the austenitic stainless steels are strongly influenced by variations in interstitially dissolved elements such as N, C, and B as shown for yield strength in Fig. 3 (ref. 11). Furthermore, equations have been developed[12] to allow relative estimates of both the room-temperature yield and ultimate tensile strengths to be made in terms of composition (wt %) and grain size as follows:

$$\text{yield strength (0.2\% offset, MPa)} \approx 15.4[4.4 + 23(\mathrm{C}) + 1.3(\mathrm{Si}) + 0.24(\mathrm{Cr}) + 0.94(\mathrm{Mo}) + 1.2(\mathrm{V}) + 0.29(\mathrm{W}) + 2.6(\mathrm{Nb}) + 1.7(\mathrm{Ti}) + 0.82(\mathrm{Al}) + 32(\mathrm{N}) + 0.16\delta + 0.46d^{-1/2}] \ , \tag{1}$$

$$\text{tensile strength (MPa)} \approx 15.4[29 + 35(\mathrm{C}) + 55(\mathrm{N}) + 2.4(\mathrm{Si}) + 0.11(\mathrm{Ni}) + 1.2(\mathrm{Mo}) + 5.0(\mathrm{Nb}) + 3.0(\mathrm{Ti}) + 1.2(\mathrm{Al}) + 0.14\delta + 0.8t^{-1/2}] \ , \tag{2}$$

where

d = grain size (mm),
t = twin spacing (mm),
δ = ferrite content (vol %).

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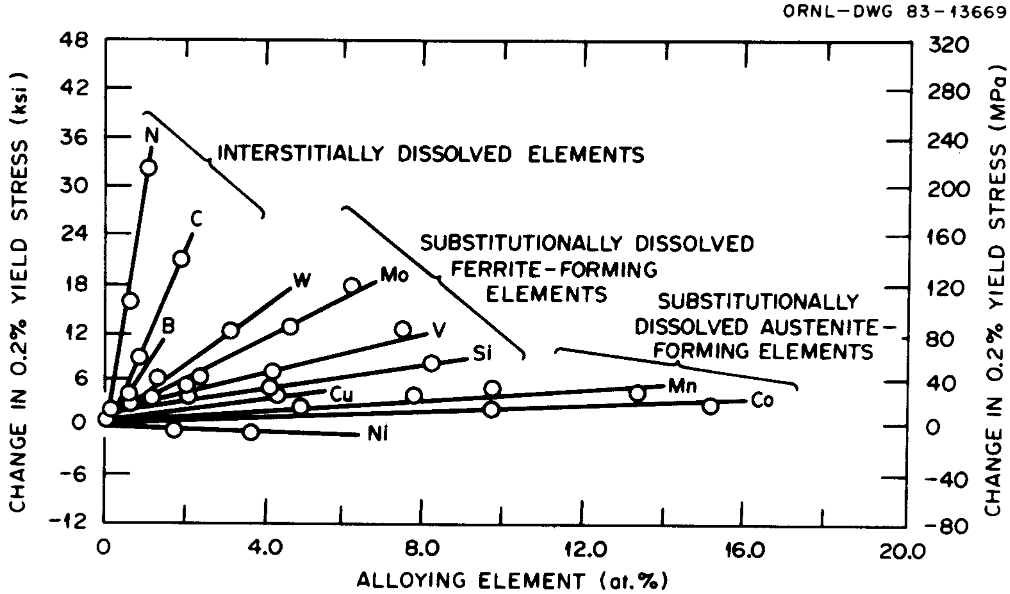

Fig. 3. Effect of solid-solution-hardening elements on the 0.2% yield stress of austenite. Based on K. J. Irvine, D. T. Llewellyn, and F. B. Pickering, "High-Strength Austenitic Stainless Steels," pp. 356–78 in *The Metallurgical Evolution of Stainless Steels*, ed. F. B. Pickering, American Society for Metals, Metals Park, Ohio, 1979.

Again, note the strong effects of the interstitial elements carbon and nitrogen on both the yield and ultimate tensile strengths.

The influence of grain size alone on the yield strength of type 304 stainless steel at both 25 and 593°C is shown in Fig. 4. The data are plotted for reannealed material in the form of the Hall-Petch relationship,

$$\text{yield strength} = \sigma_0 + kd^{-1/2} \ , \qquad (3)$$

for a number of commercial and laboratory heats in various product forms. Note the strong effects of unusually high N, C, and Nb contents on some of the yield strengths (outliers) at room temperature. High niobium contents, by formation of NbC and possibly by reduced grain size, can increase both the short-term and creep-strength properties.

In assembling the data for the various product forms plotted in Fig. 4, a rather surprising lack of data for forgings was noted. Forged

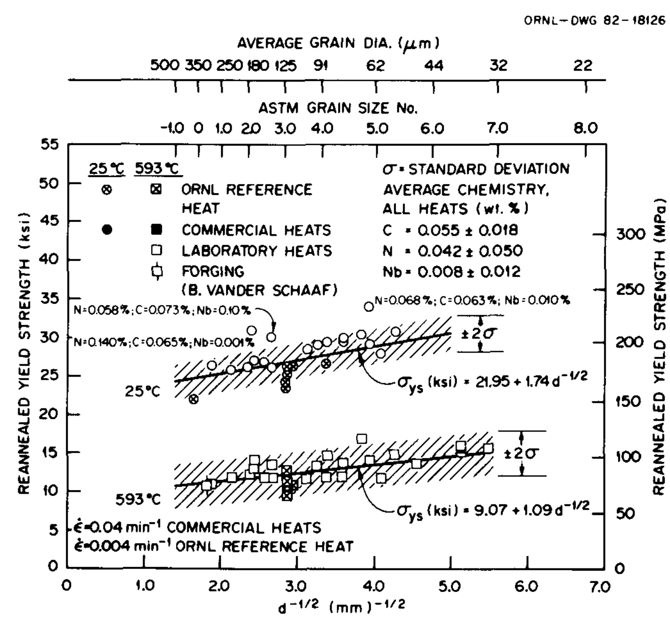

Fig. 4. Effect of grain size d on the yield strength of reannealed type 304 stainless steel at two temperatures. Forging datum from B. van der Schaaf, pp. 597–618 in *Effects of Radiation on Materials*, ASTM STP 782, American Society for Testing and Materials, Philadelphia, 1982.

materials, because of their thermomechanical processing history, have a variety of grain sizes, which can range to fairly large sizes with resultant low yield strengths, as shown by the single 593°C data point obtained from van der Schaaf.[13]

The austenitic stainless steels are generally considered to be single-phase alloys (small amounts of ferrite notwithstanding). This results in a rather low yield strength, but, because they are also face-centered cubic materials, they can be readily strengthened in terms of yield strength by cold work. Hence, it is not particularly surprising that fabricators might choose to increase the yield strengths of these steels slightly to meet specifications by minor additions of cold work by bending or straightening operations. The added cold work could lead to some elastic property variations, as will be shown, and to grain growth on subsequent solution annealing with reduced yield strengths. This appears to be the case particularly for type 304 stainless steel, as shown in Fig. 5, in which reannealed grain size is plotted as a function of mill-annealed (as-received) grain size. Note the tendency for grain sizes to increase on reannealing. Not as many data were examined for type 316 stainless steel as shown in Fig. 6, but the same trend is likely.

Figure 7 shows variations in the yield curve of several different product forms of the ORNL reference heat of type 304 stainless steel in the reannealed condition compared with various hot- or cold-drawn product forms.[14] Note the marked decrease in yield strength and the minimal variation in flow behavior after reannealing.

The combined influence of composition variations and thermomechanical processing history on the tensile properties of these steels from multiple product forms, heats, and various countries is shown in Fig. 8. Note the reduction in yield strength, with only a minor change in ultimate tensile strength on reannealing for type 304 stainless steel. Additional details concerning tensile property studies can be found elsewhere.[14–20]

Elastic constants (E, G, and ν) were measured by both static (tensile) and dynamic (sonic) methods for types 304 and 316 stainless steel.[21–23] Measurements were made on three different heats of type 304 stainless steel to determine the magnitude of any heat-to-heat variations.

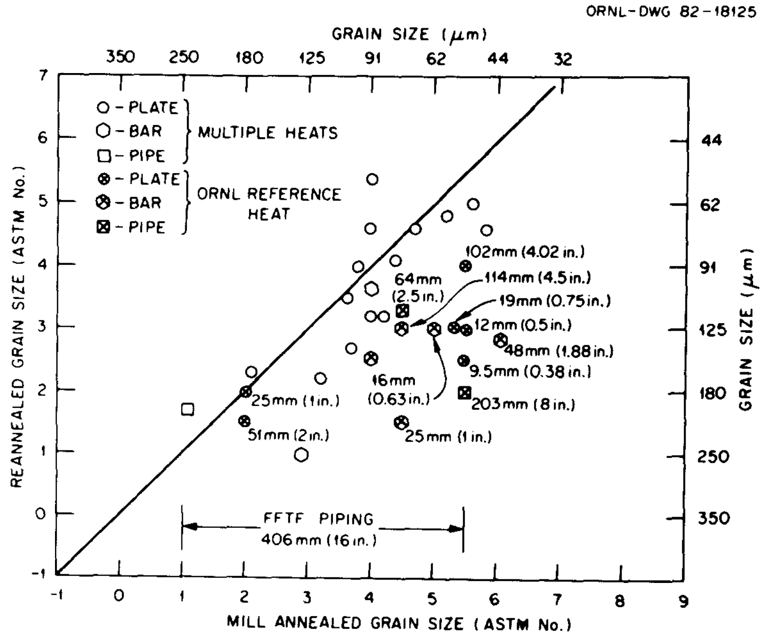

Fig. 5. Reannealing of various product forms of type 304 stainless steel decreases the ASTM grain size number (increases grain size). Grain size range for Fast Flux Test Facility piping is shown for comparison.

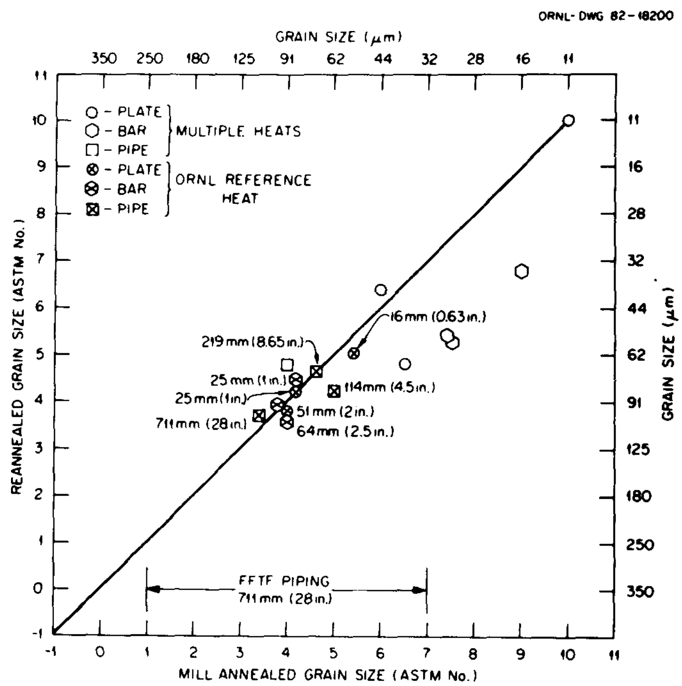

Fig. 6. Reannealing of various product forms of type 316 stainless steel may decrease the ASTM grain size number (increase grain size). Grain size range for Fast Flux Test Facility piping is shown for comparison.

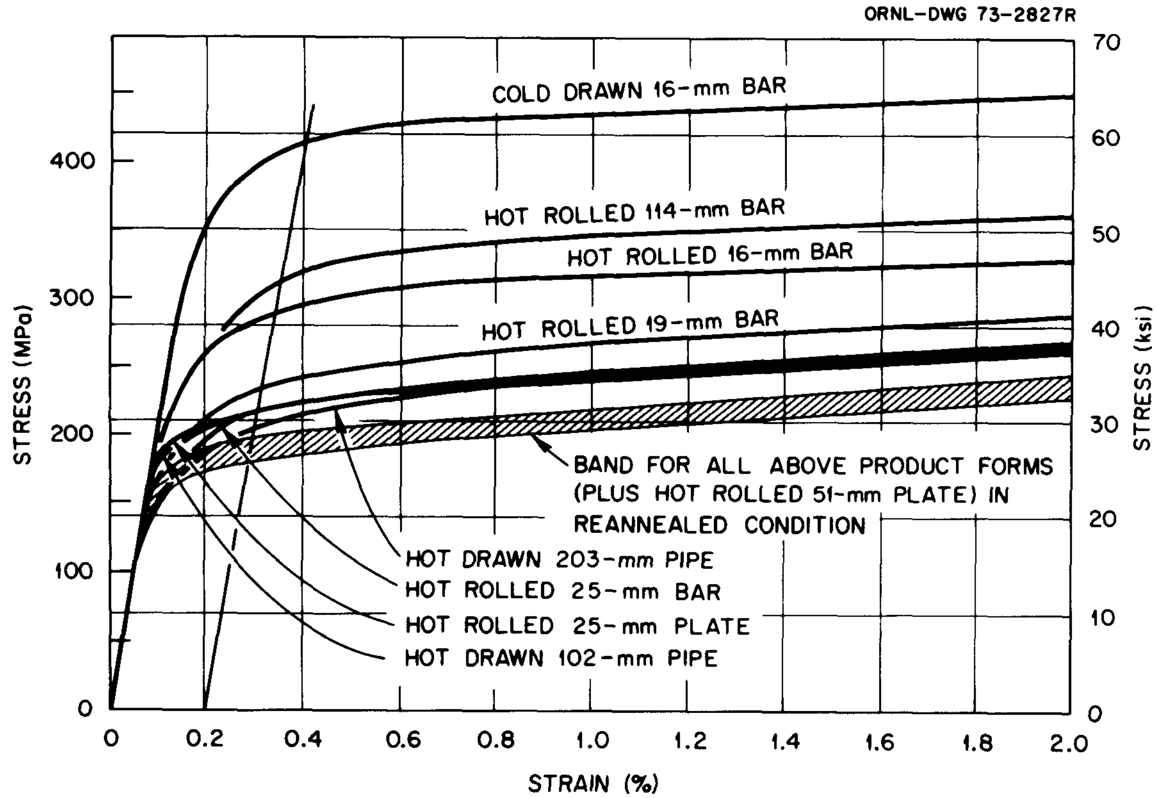

Fig. 7. Typical yield curves for several product forms of the reference heat of type 304 stainless steel.

Little difference was noted in values of Young's modulus E, but differences were found in Poisson's ratio ν, particularly at elevated temperatures. However, the differences in ν were not thought to be real but, rather, due to the techniques used for determination. It was concluded for reannealed material, which showed little anisotropy, that values of E, G, and ν varied by no more than 0.3%. However, the elastic constants of these same heats (plates) in the mill-annealed condition varied between 0.55 and 0.8%. These differences were attributed to texture introduced by some cold working in the mill straightening operations.

The heat-to-heat variability of the elastic constants in reannealed type 316 stainless steel at room temperature was also small, although it

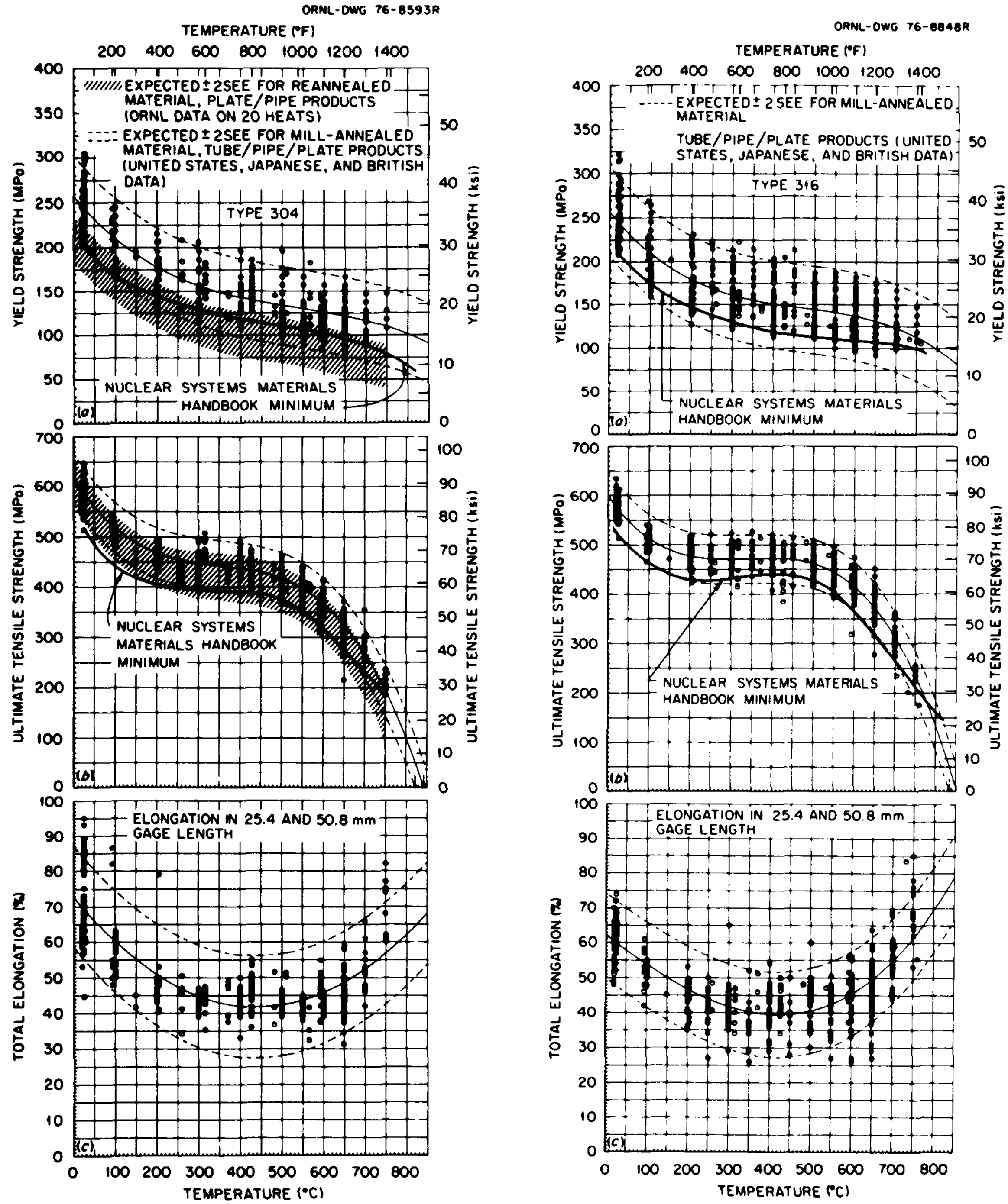

Fig. 8. Comparison of tensile properties of types 304 and 316 stainless steel.

was not as readily assessed because of effects from preferred crystallographic orientations. For four plates of thickness from 13 to 34 mm, E varied by about 0.7% and G by 1.2%. Poisson's ratio, which is more sensitive to preferred orientation, varied by 2.1%.

Types 316 and 304 stainless steel as thin plate in the reannealed condition (13 and 25 mm thick, respectively) showed significant differences in the elastic constants when measured in the different principal directions,[23] whereas 2 1/4 Cr-1 Mo steel did not, as shown in Fig. 9. For the stainless steels, E varied by about 2.2%, G by 3.2 to 3.6%, and ν by 4.8 to 5.8%. On the other hand, E, G, and ν in 2 1/4 Cr-1 Mo steel did not vary by more than 0.8%. The anisotropy in the stainless steels was associated with preferred crystallographic orientation.

Elastic constants have also been determined on as-deposited weld metal[24-25] and on castings.[26] A high degree of anisotropy due to the

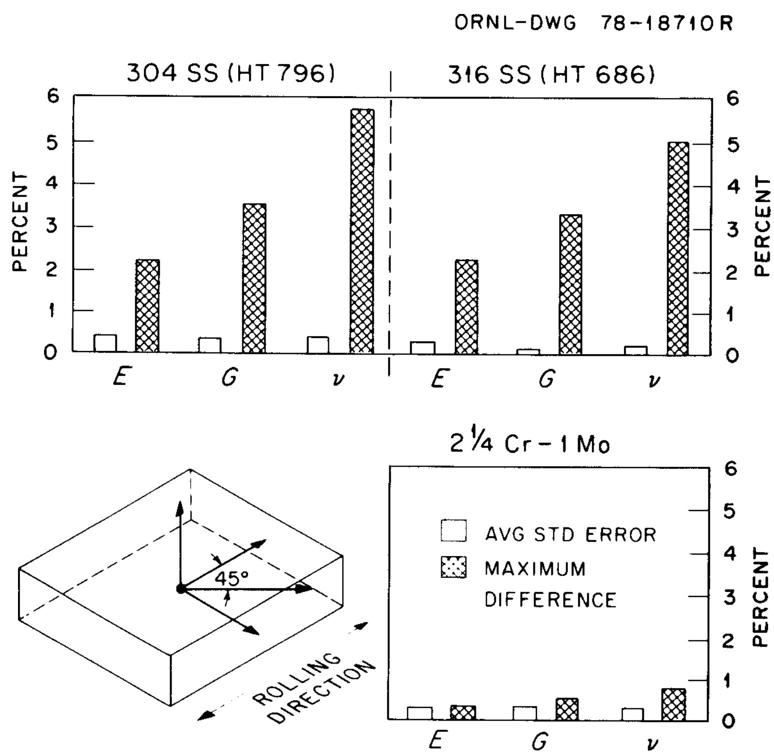

Fig. 9. Elastic constants were direction-sensitive in types 304 and 316 stainless steel but not in 2 1/4 Cr-1 Mo steel.

development of strong local preferred orientation with resultant wide variation in elastic constants was found. The designer should not assume that elastic constants are the same for both wrought and cast or weld metal.

In summary, the variations in tensile properties of these steels were demonstrated to be due to the current specifications, which permit significant variations in C, N, Nb, and B contents, and to thermomechanical processing history with resultant variations in grain size and levels of cold work.

CREEP AND CREEP RUPTURE

As indicated in Table 1, considerable variation occurs in both the creep-rupture strength and minimum creep rate and, therefore, in the predicted creep strain values for these steels. A single isothermal (593°C) rupture curve for type 304 stainless steel (Fig. 10) is indicative of the range of data scatter possible in this steel. Rupture strength values of type 316 stainless steel, a somewhat stronger steel,[19] show less variation but, as will be shown, still show considerable scatter.

Depending on the duration of the creep test, a creep test may be likened to a low-strain-rate tensile test. Therefore, it is not particularly surprising that one could look at the elevated-temperature tensile strength of a given heat in comparison with other heats of the same material for an estimate of relative creep strength. Indeed, this has been done with considerable success for type 304 stainless steel.[27–28] Thus, one would reasonably expect to receive some guidance from Eq. (1) as to the importance of various factors in determining the rupture strength of this steel. Variations in chemical elements such as C, N, Nb, B, Ti, Si, and Mo as well as in grain size would thus be considered. Figure 11 shows the importance of combined carbon and nitrogen content on 100,000-h (extrapolated) rupture strength.[29] The influence of nitrogen content on creep ductility is shown in Fig. 12 for type 316 stainless steel.[30] Figure 12 also shows that increasing the nitrogen content may increase or decrease ductility, depending on stress or minimum creep rate

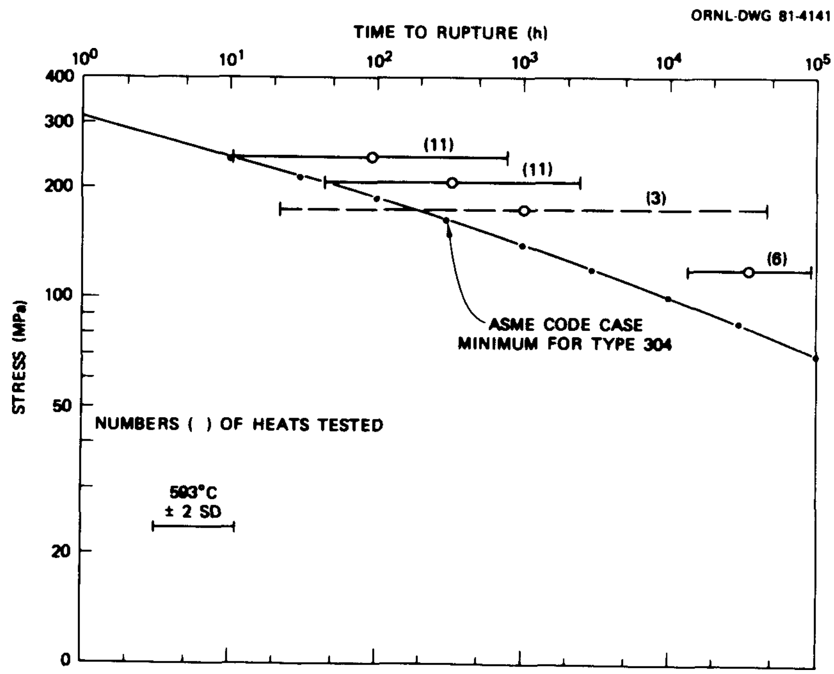

Fig. 10. Comparison of the heat-to-heat variability in rupture strength of type 304 stainless steel at several stress levels with the ASME Code Case N-47 minimum value. Data are omitted for clarity, but range or scatter in terms of plus or minus twice the standard deviation is shown.

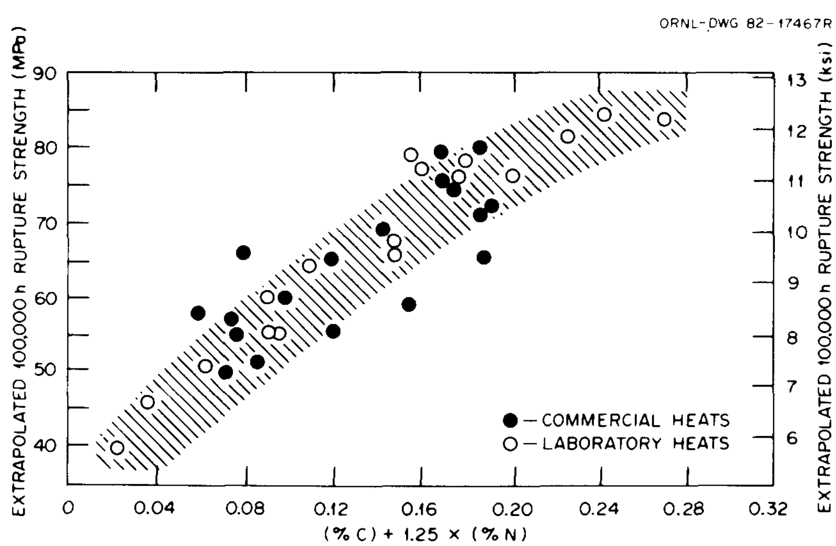

Fig. 11. Influence of carbon and nitrogen content on the extrapolated rupture strength of type 304 stainless steel. Based on data from P. D. Goodall, T. M. Cullen, and J. W. Freeman, "The Influence of Nitrogen and Certain Other Elements on the Creep-Rupture Properties of Wholly Austenitic Type 304 Steel, *J. Basic Eng.* 89, 517–24 (1967).

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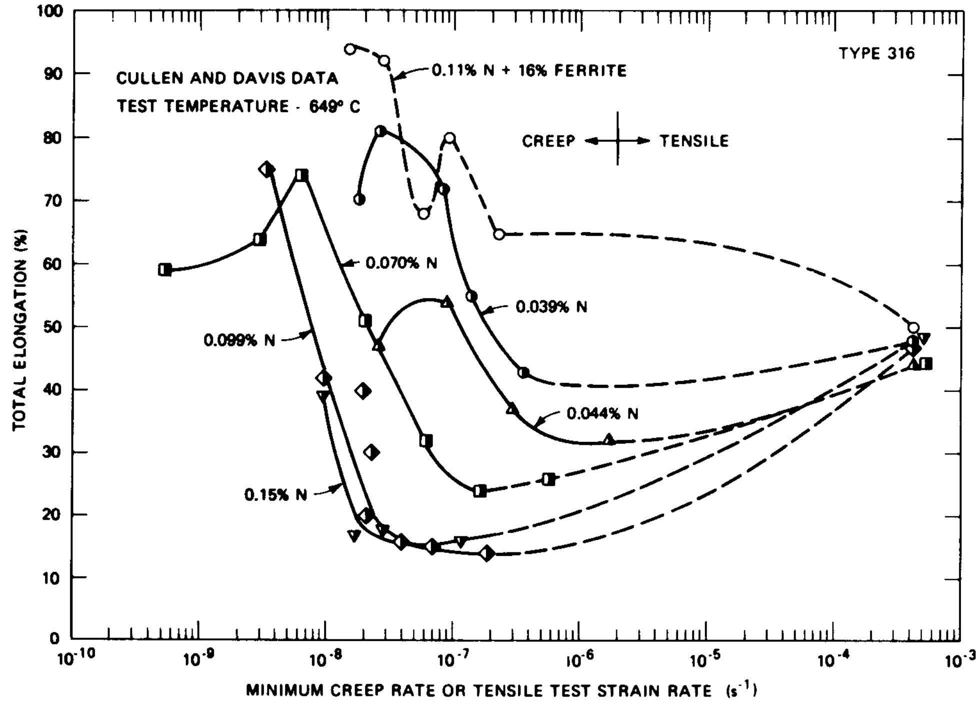

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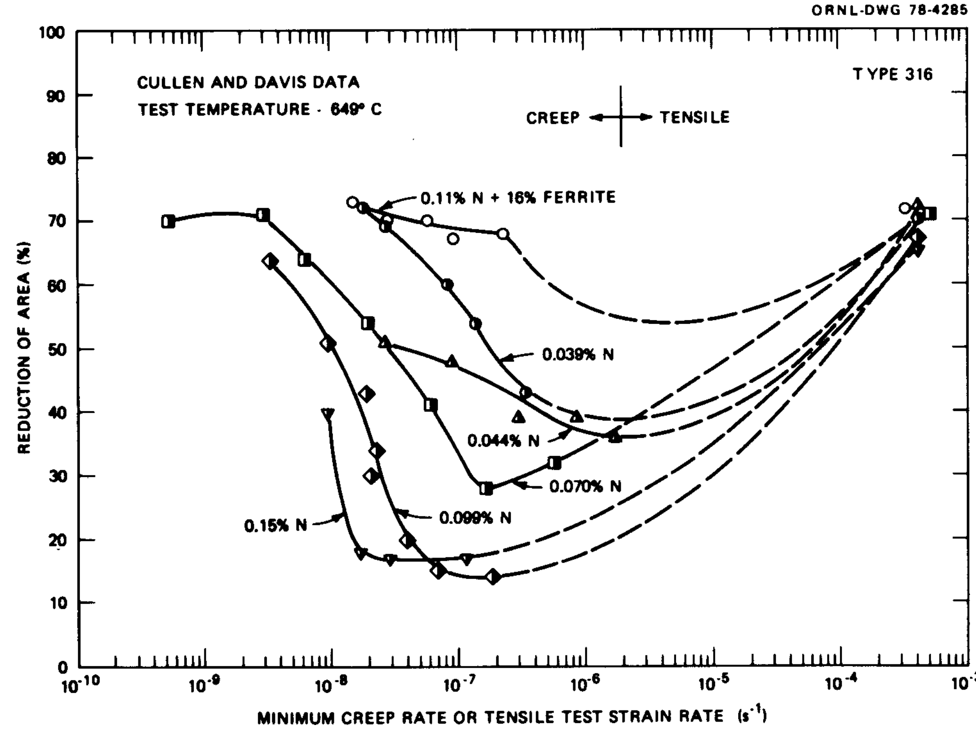

Fig. 12. Increasing the nitrogen content can increase or decrease the creep ductility depending on test duration. Based on data from T. M. Cullen and M. W. Davis, "Influence of Nitrogen on the Creep-Rupture Properties of Type 316 Steel," pp. 60–75 in ASTM STP 522, American Society for Testing and Materials, Philadelphia, 1973.

(or time at temperature because rupture time is roughly inversely proportional to minimum creep rate). Figure 12 is a good example of the complexity of material behavior when stress, time at temperature, and composition must be considered. Recently, a review of ORNL and Japanese data concluded that creep-rupture strength and ductility in these steels are largely controlled by the characteristics of the carbide or nitride phases that develop during service.[31] Thus, residual elements such as niobium,[32] boron,[7] and titanium, which can alter the kinetics of nucleation and growth of carbides or nitrides, would alter the high-temperature creep properties in a complex fashion, depending on the thermomechanical processing history and temperature of the material.

The importance of grain size in influencing the creep properties of type 304 stainless steel, at least in the near term, has been well documented.[14,31] Decreasing the grain size increases the rupture life and decreases the minimum creep rate. The latter effect can be seen in Fig. 13, in which minimum creep rate data obtained from tests[33] conducted

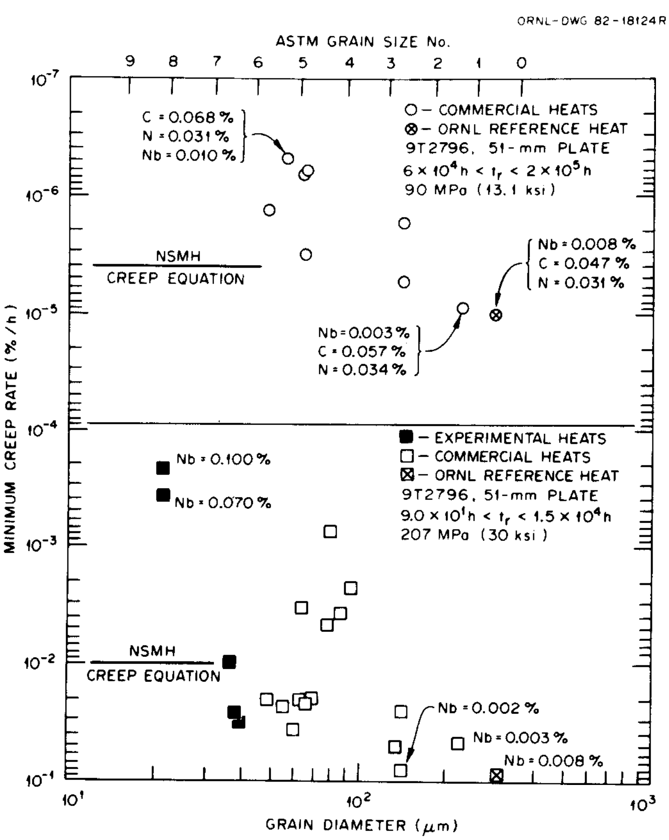

Fig. 13. Increasing grain size and decreasing niobium content result in an increase in minimum creep rate at two stress levels for type 304 stainless steel at 593°C. Note the creep scale decreases going upward. All materials shown were tested in the laboratory-reannealed condition.

at 593°C are plotted as a function of grain diameter for two stresses. The range in rupture life for the two stresses (i.e., 207 and 90 MPa) is indicated. In some instances, particularly at the lower stress, the tests were still under way as of this writing, and rupture lives and minimum creep rates were estimated from simple power law relations for individual heats.[33] Also shown is the prediction of minimum creep rate from the *Nuclear Systems Materials Handbook* (*NSMH*) equation for average-strength material. Note that the ORNL reference heat is weak (i.e., high minimum creep rate in comparison with average material). Also, note the strengthening effects (i.e., reduction in minimum creep rate) of increased amounts of niobium at the higher stress level (207 MPa).

For prolonged exposure times (i.e., 10^5 h and beyond), there is some question as to how grain size is likely to influence rupture strength of these two steels.[31] Results of long-term tests conducted at ORNL at temperatures appropriate for structural applications in fast breeder reactor service have shown, as reported above, that small grain sizes increase rupture life. Other investigators, who have generally tested these materials at somewhat higher temperatures, have reported just the opposite finding.[34] Results obtained by Clark and Freeman as reported by Smith for the 18-8 stainless steels indicated that the 10^5-h rupture strength of these steels with a grain size of 8^+ was greater than that for material with a grain size in the range from 2 to 5 up to temperatures of about 704°C (ref. 35). Above this temperature, crossover occurred, and the reverse was indicated. Creep-rupture life is likely to be a function of the relative strength of the grain boundaries and the matrix, involving the tendency to nucleate and grow voidage or cavities at the grain boundaries. In turn, this relative strength will depend on residual element content, such as that of boron, which reportedly strengthens grain boundaries and delays the onset of tertiary creep.[7] Other elements, such as carbon and nitrogen, are also likely to be important along with other factors, such as the amounts of warm or cold work present, which along with temperature would govern the amount of other phases present, such as sigma, chi, and laves.

Work at ORNL in recent years has attempted to circumvent these complex problems to some extent by use of lot-centered regression analysis for calculation of both minimum creep rate and rupture strength relationships. The method tends to integrate, over multiple test temperatures, the complex material variables for a given heat. Thus, assuming a reasonable-size data base, the method gives expressions for rupture times and minimum creep rates in terms of temperature, stress level, and a lot constant, which is indicative of the strength of a particular heat or product form of a heat. The equations in their present form follow.[36-37]

Rupture Life

Type 304 stainless steel:

$$\log t_r = C_h - 0.01258T - 1.232 \times 10^{-5} T\sigma - 0.004770T \log \sigma \ . \tag{4}$$

Type 316 stainless steel:

$$\log t_r = C_h - 0.01312\sigma - 2.552 \log \sigma + 20{,}880/T \ . \tag{5}$$

In the above equations all logarithms are base 10, t_r is the rupture life (h), σ is the stress (MPa), and T is the temperature (K).

Minimum Creep Rate

Type 304 stainless steel:

$$\log \dot{e}_m = C_h' + 0.02113T + 1.754 \times 10^{-5} T\sigma + 5.218 \times 10^{-3} T \log \sigma \ , \tag{6}$$

and type 316 stainless steel:

$$\log \dot{e}_m = C_h' + 0.0097\sigma + 4.5097 \log \sigma - 24{,}890/T \ , \tag{7}$$

where $\dot{e}_m$ is the minimum creep rate (%/h), and the other variables are as defined above. The lot or heat constants C_h and C_h' reflect the relative strengths of different heats, as shown in the lot constants listed below.

	Type 304 stainless steel	Type 316 stainless steel
Low strength		
Rupture strength	24.266	−12.674
Minimum creep rate	−32.291	14.293
Average strength		
Rupture strength	25.458	−11.870
Minimum creep rate	−33.890	13.314
High strength		
Rupture strength	26.650	−11.065
Minimum creep rate	−35.488	12.335

Low- and high-strength values of the lot constants were calculated by simply adding or subtracting twice the overall standard error of estimate to the average-strength values of lot constants given above.

In Fig. 14 values of stress-rupture life and minimum creep rate are plotted for relatively weak and strong heats at several stress levels and temperatures. The values shown were calculated with Eqs. (5) and (7) for two actual heats of type 316 stainless steel.[37] Use of the two equations in this fashion shows an inverse relationship between time to rupture and minimum creep rate, which, when plotted on log-log coordinates, tends to approach a linear relationship for a given temperature. A direct plot of experimental data for 19 heats of type 304 stainless steel tested at 593°C is given in Fig. 15. A power law relationship between minimum creep rate and time to rupture appears to be appropriate when all the data are viewed collectively. This simple power law is a form of the Monkman-Grant equation.[38] Use of the Monkman-Grant correlation for the reference heat of type 304 stainless steel has been shown to be invalid as a result of both the known temperature dependence and the departure from linearity (log-log coordinates) known to exist at low stress levels.[39] However, the intent of Fig. 14 was to display graphically minimum creep rate and time to rupture (solid lines) ranges that can be calculated for a given stress

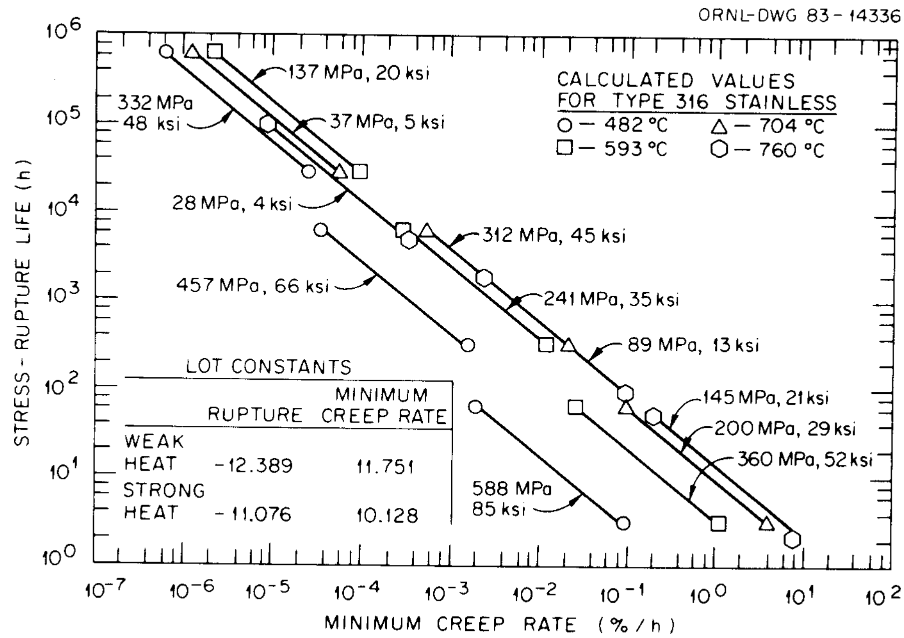

Fig. 14. Inverse relationship between stress-rupture life and minimum creep rate for type 316 stainless steel. Calculated ranges in values between an actual strong heat and an actual weak heat are given for several temperatures and stresses.

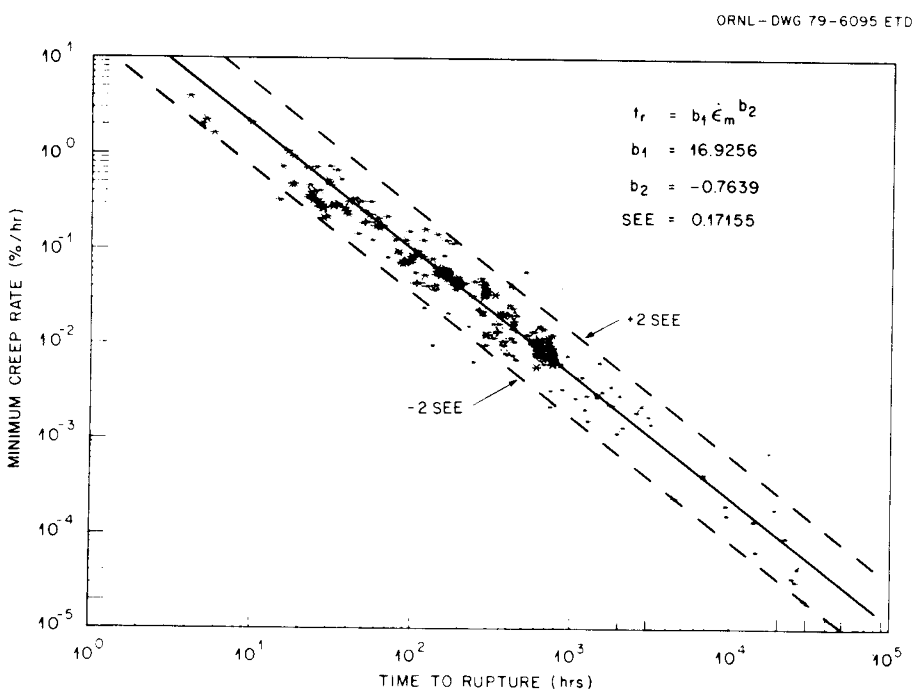

Fig. 15. Correlation between time to rupture and minimum creep rate at 593°C (1100°C) for 19 heats of annealed type 304 stainless steel. *Source*: W. J. McAfee and W. K. Sartory, *Materials Heat-to-Heat Variability Study: Part 1 — Compilation and Analysis of Data*, ORNL-5604, November 1980.

and temperature from lot constants for two actual heats of this material. We concluded that, without a knowledge of the specific lot constants for a given heat of material, the best that can be done in creep-related calculations of this type is to simply bracket expected behavior.

The rupture lot constants have been plotted as a function of ultimate tensile strength for types 304 and 316 stainless steel in Fig. 16. Note that for type 304 stainless steel a clear relationship appears to exist between ultimate tensile strength and the lot constant. By definition, this means that strong heats, as measured by simple elevated-temperature tensile tests, remain strong relative to weak heats during prolonged exposure to creep. Indeed, analysis of ORNL data suggests that this is true. Comparison of long-term Japanese creep data with failure times in the range from 10^3 to 10^5 h also supports this conclusion.[31] Figure 17 shows the creep-rupture strength of several heats of type 304 stainless steel currently under test at 593°C, and Table 6 gives pertinent properties of these heats. The lines in Fig. 17 indicate predicted rupture life for the heats shown as obtained from Eq. (4). All the data shown were used in formulating Eq. (4) except the single test result shown for heat 414 at 94,500 h. Note that good agreement was obtained between experimental and analytical results and that a decrease in grain size generally results in an increase in the lot constant, with resultant increased stress to failure for a given time. The single exception to this trend is heat 414, which has the highest lot constant, probably because of the strengthening influence of both carbon and nitrogen in conjunction with a below average grain size. In the case of type 316 stainless steel, Fig. 16(*b*) shows that the trend is not as clear. In fact, exceptions are known to the concept that weak heats remain weak and strong heats remain strong.[7] Figure 18 is a plot of British rupture data showing examples of crossover (i.e., heats of intermediate strength becoming stronger with time).[7] Other examples can be found in the literature.[7] The data shown in Fig. 18 were not included in the analysis used to derive the lot constants for type 316 stainless steel and are therefore supportive of the information given in Fig. 16(*b*); namely, that the creep behavior of type 316 stainless steel, although showing less data scatter than that of type 304, is more difficult to relate to short-term properties.

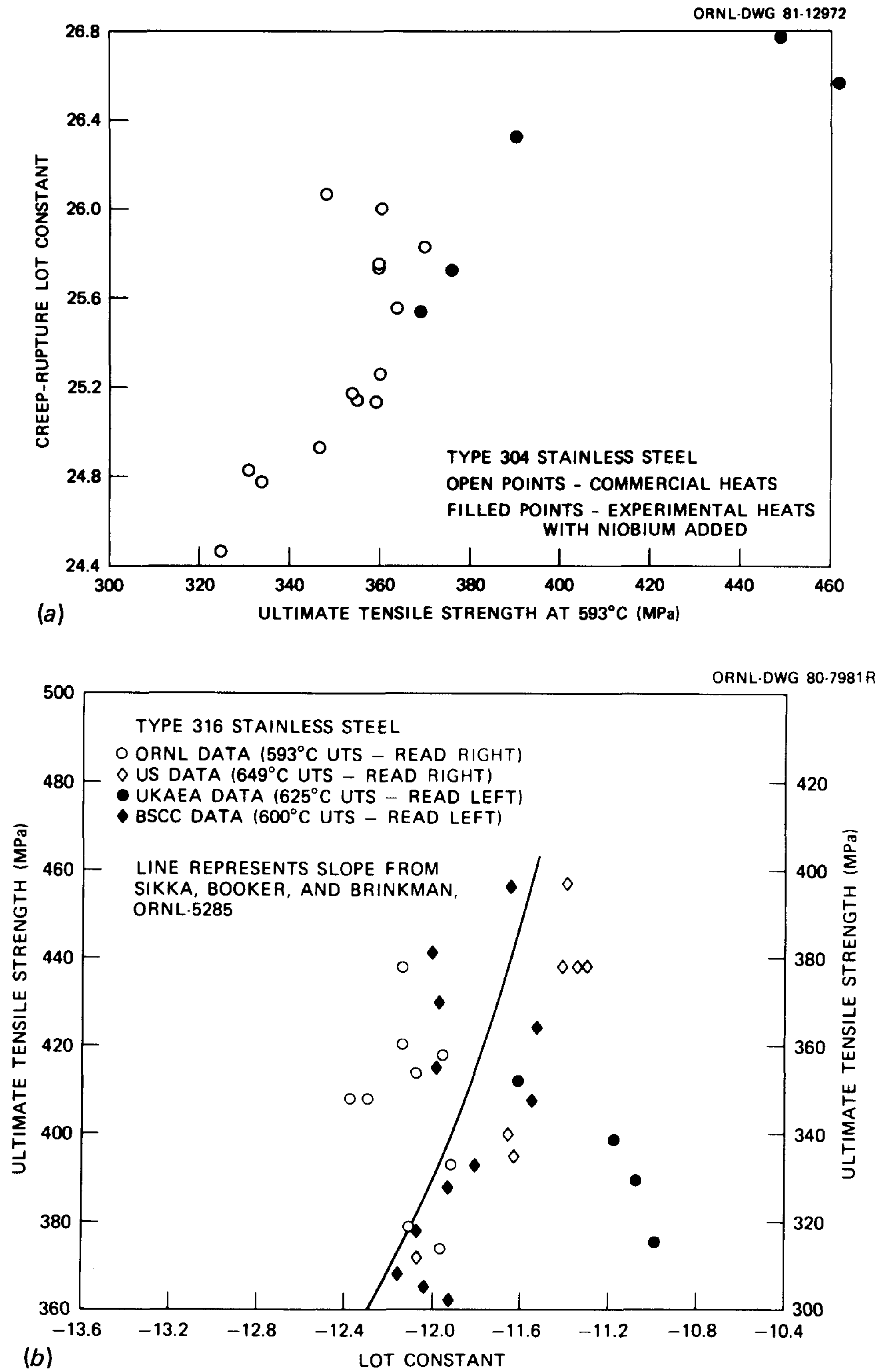

Fig. 16. Rupture lot constants as a function of ultimate strength. (*a*) Type 304 stainless steel. (*b*) Type 316 stainless steel.

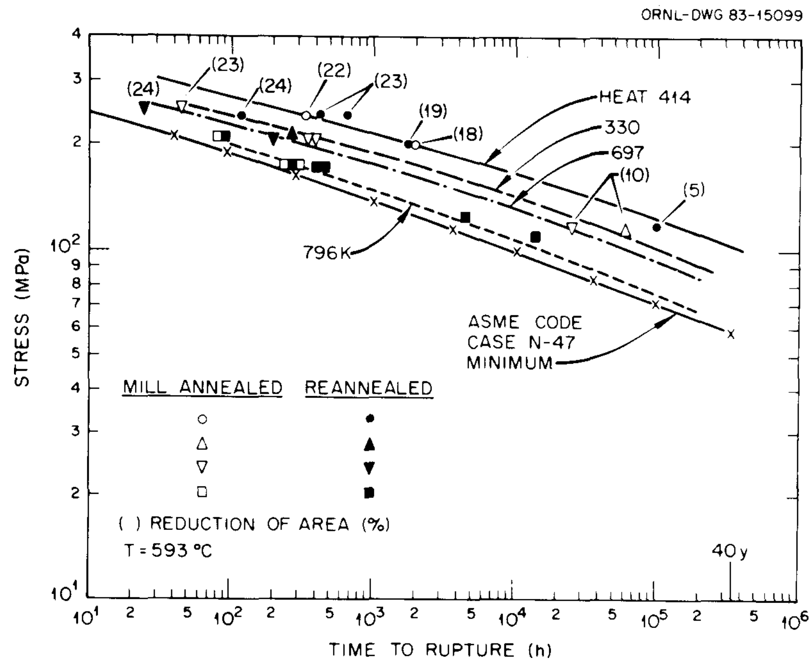

Fig. 17. Effect of grain size and composition on the rupture strength (lot constant) of type 304 stainless steel at 593°C. Note that a small grain size and a high combined carbon and nitrogen content lead to improved rupture strength.

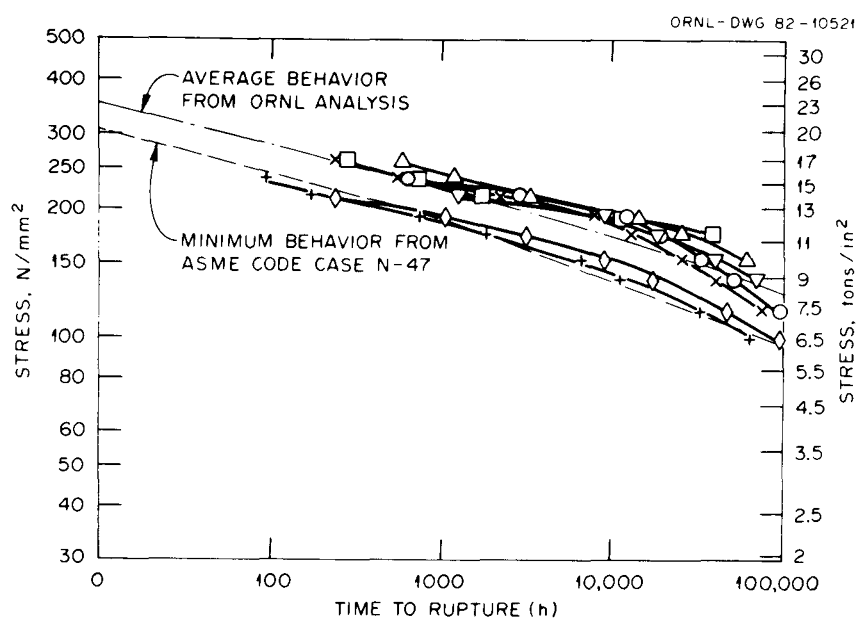

Fig. 18. Typical results from United Kingdom national stress-rupture testing program. British long-term creep-rupture data tend to verify ASME Code predictions for type 316 stainless steel at 600°C. Data from H. K. Grover and A. Wickens, pp. 81–87 in *Mechanical Behavior and Nuclear Applications of Stainless Steel at Elevated Temperatures*, The Metals Society, London, 1981.

Table 6. Selected properties of type 304 stainless steel heats represented in Fig. 17

Symbol in Fig. 17	Heat	Lot constant	Grain size[a]		Product form and thickness (mm)	Content (wt %)		
			ASTM	(μm)		C	N	Nb
△ ▲	330	25.261	5.6/5.0	45/55	Plate (29)	0.07	0.03	0.01
○ ●	414	25.827	3.7/2.7	97/140	Plate (60)	0.07	0.06	0.01
▽ ▼	697	25.044	2.9/1.0	115/220	Bar (16)	0.06	0.03	0.003
□ ■	796K	24.466	0.26/0.26	280/280	Plate (51)	0.05	0.03	0.008

[a]Mill-annealed/reannealed.

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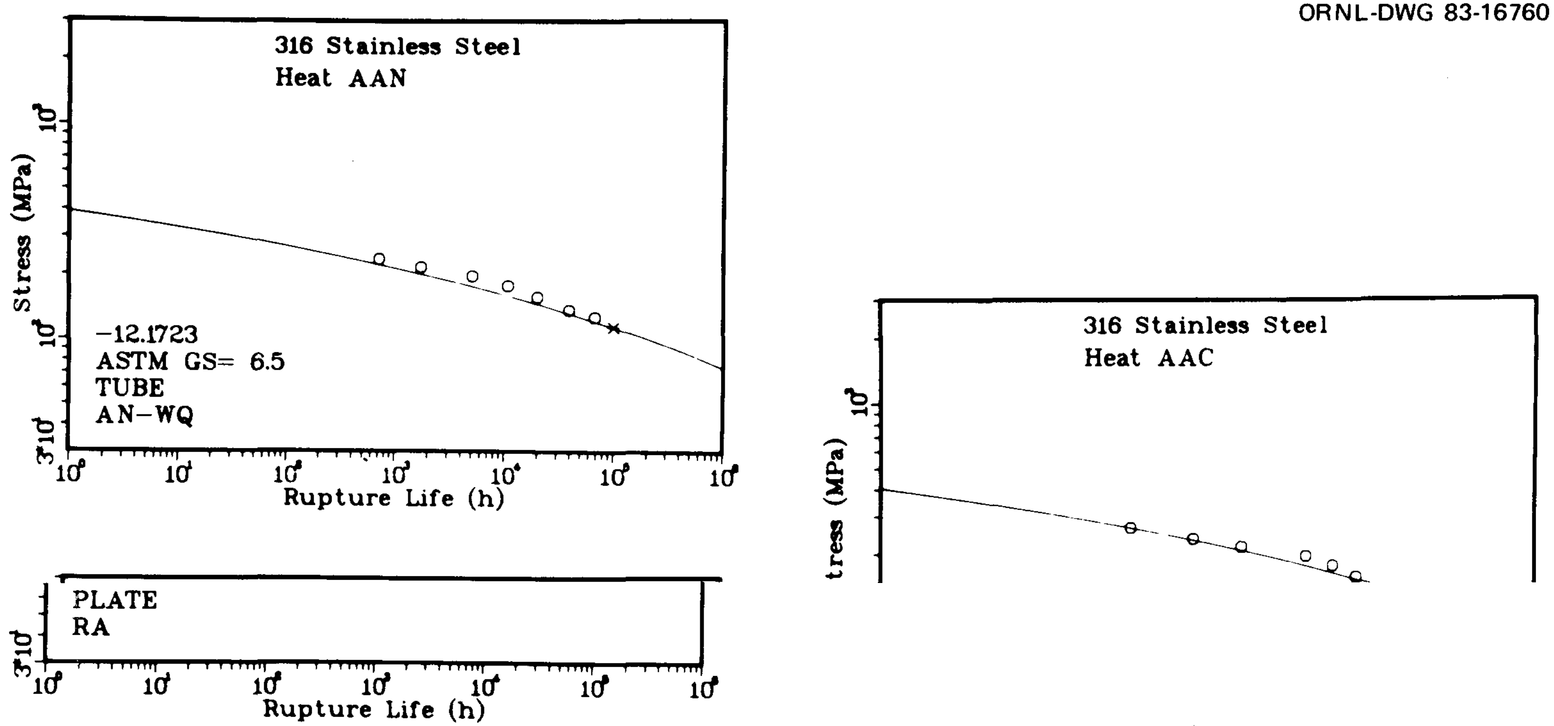


Fig. 19. Comparison of estimated 10^5-h rupture life (*) and creep-rupture curve (solid line) with actual data (o) from tests conducted at 593 to 600°C. The "apparent" lot constant, ASTM grain size, product form, and heat treatment are listed in the lower left for each heat tested. Note that a visual comparison indicates that the method of estimating rupture life is most accurate when the "apparent" lot constant is within 0.3 of average lot constant (−11.870) calculated from actual rupture data.

British investigators, using regression analysis and the data from long-term creep-rupture tests on type 316 stainless steel, have formulated several equations to predict the 10^5-h rupture life of this steel.[7] They conclude that long-term rupture strength can be predicted within 12% with the following equations in terms of composition (wt %):

$$\begin{aligned}\text{stress to rupture (MPa)}&\\ \text{(pipes and bars at 600°C)} &= [10.2 \times 0.926(\%\ \mathrm{Mn}) - 0.635(\%\ \mathrm{Ni})\\ &\quad -2.91(\%\ \mathrm{Si}) + 72.6(\%\ \mathrm{S}) + 77.4(\%\ \mathrm{N})\\ &\quad + 523(\%\ \mathrm{B})] \times 15.443\ , \end{aligned} \tag{8}$$

$$\begin{aligned}\text{stress to rupture (MPa)}&\\ \text{(pipes and bars at 625°C)} &= [5.36 - 3.53(\%\ \mathrm{Si}) + 43.0(\%\ \mathrm{N})\\ &\quad + 251(\%\ \mathrm{B})] \times 15.443\ . \end{aligned} \tag{9}$$

Note the importance of both nitrogen and boron and how it changes with temperature. Also of interest is the apparent strengthening effect of sulfur at 600°C. This effect of sulfur was not observed in type 304 stainless steel.[31] Also note that Eqs. (8) and (9) are independent of carbon concentration, yet the British investigators concluded that the optimum carbon concentration was about 0.05%. This observation no doubt influenced the decision of United Kingdom Atomic Energy Authority engineers in setting the more restrictive carbon limitations of 0.03 to 0.06% in the revised CDFR-CFRX specifications given in Table 4.

As a further test of the above equations, we compared the calculated 10^5-h rupture life with independent data obtained from U.S. and Japanese sources. Data obtained from tests conducted at 600°C on about 30 different heats or product forms were included in this analysis. Also the "apparent" lot constant from the 10^5-h calculated rupture life was determined for each heat from Eq. (5). Thus, the entire calculated creep-rupture curve could be estimated in this fashion and compared with the actual data. Figure 19 compares the results of this analysis. Note that the method (according to visual inspection) produces a satisfactory estimate of rupture life along the entire curve where data are available when the apparent lot constant is close to the average lot constant of −11.870

calculated from actual data.[37] When the apparent lot constant is more negative than −11.870, the method predicts a lower creep-rupture strength for that heat or product form than what was observed. When the apparent lot constant is greater than average (e.g., −10.9116 as for heat 8092297), the method predicts a stronger response than what in fact was observed. Overall, the method shows promise but needs further elaboration as more data are examined, particularly at other temperatures.

Table 7 qualitatively compares the apparent influence of an increase in grain size or individual chemical constituent on the rupture strength of types 304 and 316 stainless steel. In this analysis, the influence of an individual metallurgical variable on the stress-rupture lot constant was examined as shown in the appendix. Table 7 summarizes the conclusions drawn from the comparisons given in the appendix and compares them with conclusions drawn over 30 years ago by Smith[35] and more recently by Grover and Wickens[7] for type 316 stainless steel. As indicated above, Grover and Wickens used multiple regression analysis to examine the effects of all elements simultaneously, rather than individually as was done in the ORNL analysis given in the appendix. A multiple regression analysis of all data collected at ORNL has been deferred pending the completion of ongoing long-term tests. The comparisons given in Table 7 indicate that reduced grain size increases rupture strength, as does an increase in Mo, C, N, Nb, and B contents. Elements such as phosphorus and silicon may be either strengthening or weakening, depending apparently on test duration, temperature, and other constituents present. It should be again emphasized that the strengthening effect of these elements is not necessarily due to individual interaction but rather to a synergistic interaction, which can change with time, temperature, and perhaps loading conditions.

Another question that logically could be asked regarding the ability to predict long-term creep and creep-rupture response from composition concerns the influence of small levels of cold work (i.e., mill-annealed versus reannealed material response). As mentioned in the previous section, mill-annealed material usually contains a small amount of cold

Table 7. Qualitative influence of grain size and chemistry on the rupture strengths of types 304 and 316 stainless steel

Increase in material variable	Influence on rupture strength			
	Current analysis		G. V. Smith[a] (1950)	Grover and Wickens[b] type 316 (1981)
	Type 304	Type 316		
Grain size	Pronounced reduction	Slight reduction	Reduction	
Mo	Increase	Increase	Increase	
C	Increase	Increase	Increase	Increase[c]
N	Increase	Increase		Inconclusive
Nb	Pronounced increase		Increase	
B	Slight increase	Pronounced increase		Increase
Mn	Inconclusive	Inconclusive	Inconclusive small effect	
Si	Apparent increase	Apparent increase	Small effect	Decrease[d]
S	Decrease	Inconclusive		
Cr	Slight decrease	Increase	Small effect	
Ni	Slight decrease	Decrease	Inconclusive	
P	Slight decrease	Increase		Increase[e]

[a]*Properties of Metals at Elevated Temperatures*, McGraw-Hill, New York, 1950, p. 292.

[b]H. K. Grover and A. Wickens, "Effects of Minor Elements on the Long-Term Creep Rupture Properties of Type 316 Austenitic Stainless Steel in the Range 500–700°C," pp. 81–87 in *Mechanical Behavior and Nuclear Applications of Stainless Steel at Elevated Temperatures*, The Metals Society, London, 1981.

[c]Carbon contents varied from 0.038 to 0.07%, the highest strength was with heats containing above 0.05% C at 600 to 650°C.

[d]Recommended silicon be maintained below 0.2% to reduce scatter and maintain strength.

[e]Strengthening to 50,000 h at 600°C; at higher temperature and longer times, it is apparently weakening.

work due to straightening operations. The amount of cold work depends on the product form; our experience indicates that bars contain more (~10%) residual cold work than do pipe or plate (3–4%), but this is likely to depend on thickness or diameter.[40] Forgings are likely to have a wide range of grain sizes, and possibly some as-cast microstructure present may confuse the issue. An extensive comparison of the creep and creep-rupture behavior of mill-annealed versus reannealed type 304 stainless steel is an ongoing part of the current creep-testing program at ORNL. These tests[33] have been conducted at 593°C on several heats and product forms (plate and bar) of differing thickness. In the near term, the small amount of cold work clearly increases yield strength (Figs. 7 and 8), slightly increases rupture life, and decreases minimum creep rate in comparison with reannealed materials.[40] This is what one would expect from increased strength and smaller grain sizes (Figs. 4 and 13). In the long term, the rupture lives and minimum creep rates of the mill-annealed material tend to approach those of the reannealed material for tests conducted at 593°C. Differences in behavior will of course be maintained for longer times at lower temperatures.

In summary, we concluded the following.

1. Many of the same variables that impact short-term ultimate tensile strength values, such as composition and grain size, are also important in determining the creep-rupture and minimum creep rate of type 304 stainless steel. Hence, grain size control and some limitations on variability in contents of such elements as Nb, Ti, Mo, B, C, and N, when practical, should be considered to minimize data scatter. The current success in predicting both the creep and creep-rupture response of both types 304 and 316 stainless steel with lot-centered analysis suggests the value of a short-term creep test as a means of estimating the appropriate lot constants when calculations of creep deformation are required.

2. In estimating the creep and creep-rupture response of type 316 stainless steel, the relationship to short-term tensile strength is not well defined. Available data from the literature suggest that, from 593 to 625°C, control of carbon content to about 0.05% and of other elements

such as B, N, and perhaps S can reduce data scatter. Details of composition control will be explored in greater detail in the discussion section.

3. Currently available data suggest that small levels of cold work normally associated with certain product forms of mill-annealed material make only a small contribution (if any) to the long-term creep response of type 304 stainless steel at temperatures around 593°C. At higher levels of cold work (e.g., 10–30%), significant strengthening will occur in the near term; however, if the temperatures are high enough, recrystallization with resultant decreased rupture life may occur.[41]

4. Control of material composition specifications at the residual element level in both steels is confused by the fact that certain elements acting in combination with other elements may be strengthening or weakening on prolonged exposure to creep at elevated temperatures.

CONTINUOUS-CYCLE FULLY REVERSED FATIGUE

Representative elevated-temperature strain-controlled fatigue data for types 304 and 316 stainless steel are shown in Figs. 20 and 21, respectively.[20] These data represent by no means all the data available but are indicative of general trends to be discussed. Characteristically, strain-controlled continuous-cycle data variations tend to be sensitive to major ductility variations in the low-cycle regime and are sensitive to variations in strength in the high-cycle end of the curve. Figure 20 shows variations observed in eight different heats of type 304 stainless steel following several heat treatments when tested by three different laboratories. Also shown is the range in data obtained by one laboratory when testing one of these heats at a single strain range ($\Delta\varepsilon_t$ = 0.5%) with zero hold time.[42] The material was initially in the solution-annealed and aged condition. Note that the life ranged from 27,552 to 73,318 cycles to failure. Figure 21 presents results from tests conducted by three laboratories on a single heat of type 316 stainless steel. Again the variation in data scatter is shown at a total strain range of 0.5%. Note the order of magnitude in scatter produced by differences in heat treatment, differences in specimen geometry (hourglass diameter, 5.08 and

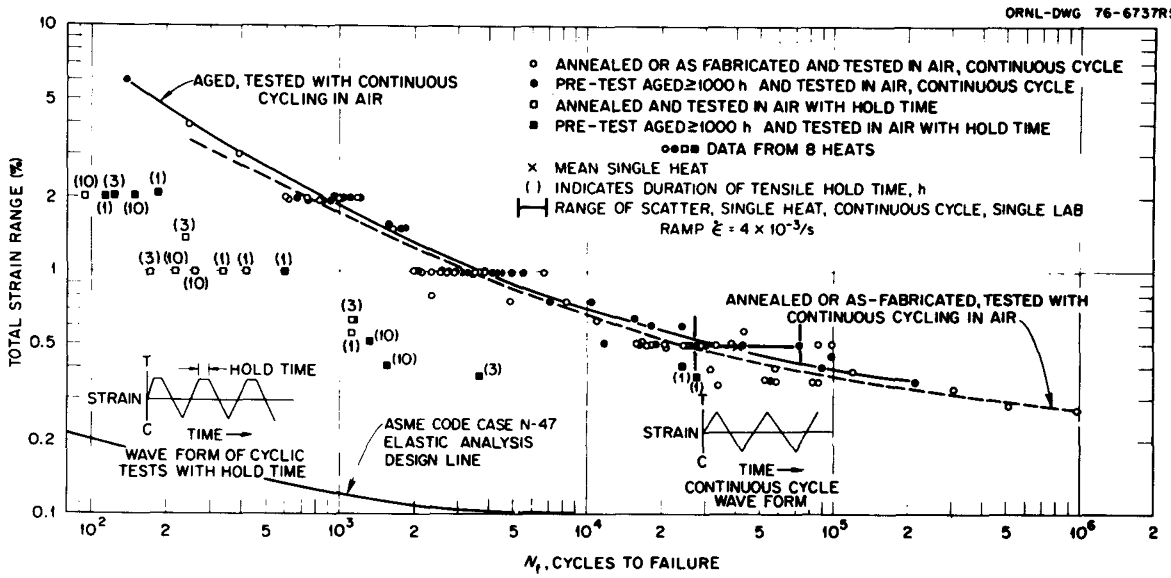


Fig. 20. Comparison of strain-controlled fatigue data on type 304 stainless steel at 538 to 593°C in air.

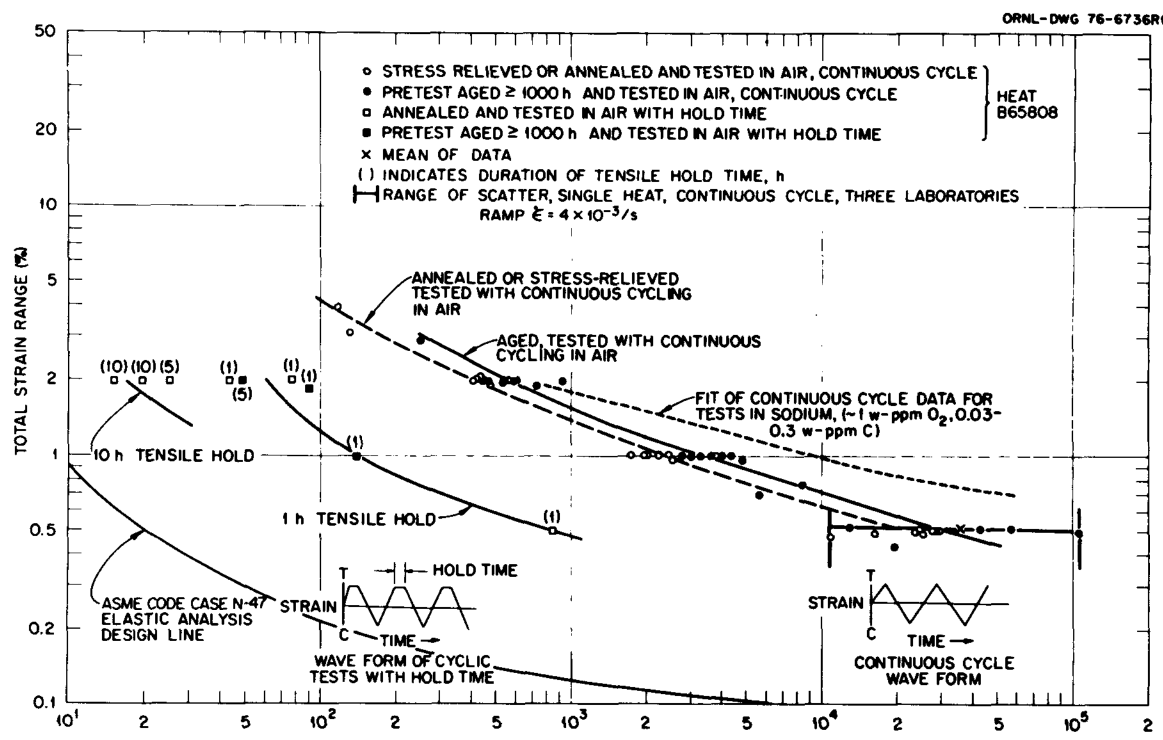


Fig. 21. Comparison of strain-controlled fatigue data on type 316 stainless steel at 566 to 593°C in air or sodium.

6.35 mm), and subtle differences in test technique (specimen rigidity and alignment) introduced by the three laboratories.[43] Continuous-cycle data plotted in Figs. 20 and 21 show that thermal aging before testing (precipitation of $M_{23}C_6$) results in a modest increase in fatigue life.

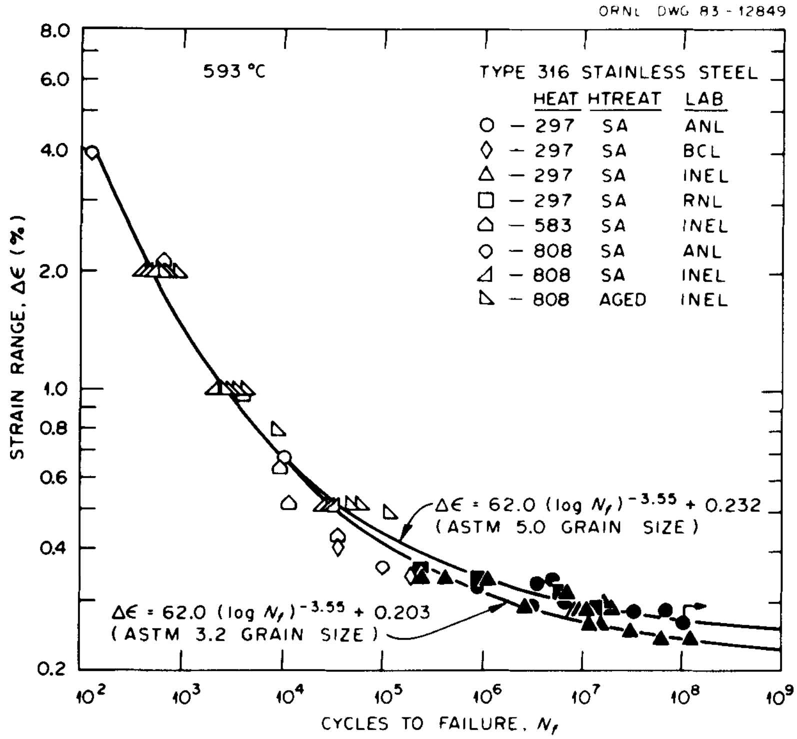
Recently, Raske and Korth have clearly demonstrated that part of the apparently inherent high data scatter found in the high-cycle end of the fully reversed fatigue curve can be due to grain size differences.[44] This is demonstrated for type 316 stainless steel in Fig. 22. Note that increasing the grain size (decreasing ASTM number) reduces the fatigue life.

In summary, we conclude that, at strain ranges of 0.5 or less (the range of most interest to the designer), data scatter in types 304 and 316 stainless steel fully reversed fatigue data is due to grain size differences, to interlaboratory variations in test techniques, and to the metallurgical instability of these steels at high temperature. To increase high-cycle fatigue life, the data indicate that grain size should be minimized. Thermal aging may slightly decrease the continuous-cycle fatigue life at lower temperatures as a result of carbide cracking, particularly at high strain ranges.[45]

CREEP-FATIGUE INTERACTION (STRAIN CONTROL)

Figures 20 and 21 contain some of the strain-controlled hold time data available for types 304 and 316 stainless steel. Much of the data currently available has been omitted for clarity. However, sufficient data are plotted to show that metallurgical state is an important parameter and that thermal aging can have a pronounced beneficial effect on the fatigue life of these materials when tensile hold periods are imposed.

One of the first studies dealing with heat-to-heat variations in the creep-fatigue behavior of type 304 stainless steel compared the creep-fatigue response of five heats, as shown in Fig. 23 (ref. 46). Specific details of heat composition, grain size, and stress-rupture lot constants[36] [C_h of Eq. (4)] are given in Table 8. The data given in Fig. 23 indicate that one of the five heats (8043813) has superior resistance to time-dependent fatigue (higher fatigue life for a given



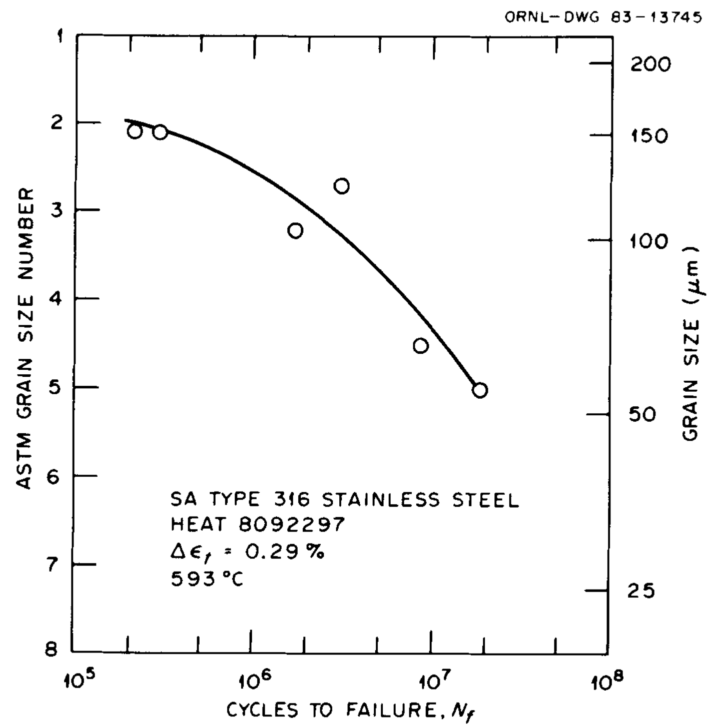


Fig. 22. Decreasing the grain size increases the high-cycle fatigue life of type 316 stainless steel at 593°C. *Source:* D. T. Raske and G. E. Korth, *Elevated-Temperature, High-Cycle Fatigue Behavior of Type 316 Stainless Steel*, ANL-83-35, Argonne National Laboratory, Argonne, Ill., July 1983.

Table 8. Chemical composition, grain size, and lot constants for several heats of type 304 stainless steel

Heat[a]	Content (wt %)[b]												Product form and thickness (mm)	Reannealed grain size		Stress-rupture lot constant[b,c]
	C	N	P	Ni	Mn	Cr	Si	Mo	S	Nb	V	Ti		ASTM	(μm)	
380	0.063	0.068	0.018	8.30	0.97	18.4	0.55	0.07	0.01	0.01	0.028		60, Plate	4.6	62	25.539
414	0.073	0.058	0.016	9.52	0.94	18.7	0.69	0.10	0.015	0.01	0.025	0.002	60, Plate	2.7	140	25.827
697	0.057	0.034	0.016	9.38	0.91	18.5	0.50	0.05	0.037	0.003	0.03	0.002	15.9, Bar	1.0	220	25.044
796	0.048	0.031	0.028	9.7	1.22	18.6	0.48	0.32	0.015				Bar	1–3	220–130	24.683[d]
813[e]	0.062	0.033	0.044	8.95	1.87	17.8	0.48	0.32	0.004	0.02	0.022	0.002	25.4, Plate	4.0	78	25.8

[a]Shortened to last three digits.

[b]Data from M. K. Booker, V. K. Sikka, and B. L. P. Booker, *Analysis of the Creep Strain-Time Behavior of Type 304 Stainless Steel*, ORNL-5836, February 1982.

[c]C_h in $\log t_r = C_h - 0.01258T - 1.232 \times 10^{-5} T\sigma - 0.004770T \log \sigma$.

[d]Lot constant given for plate material.

[e]Contains 3 wt ppm B.

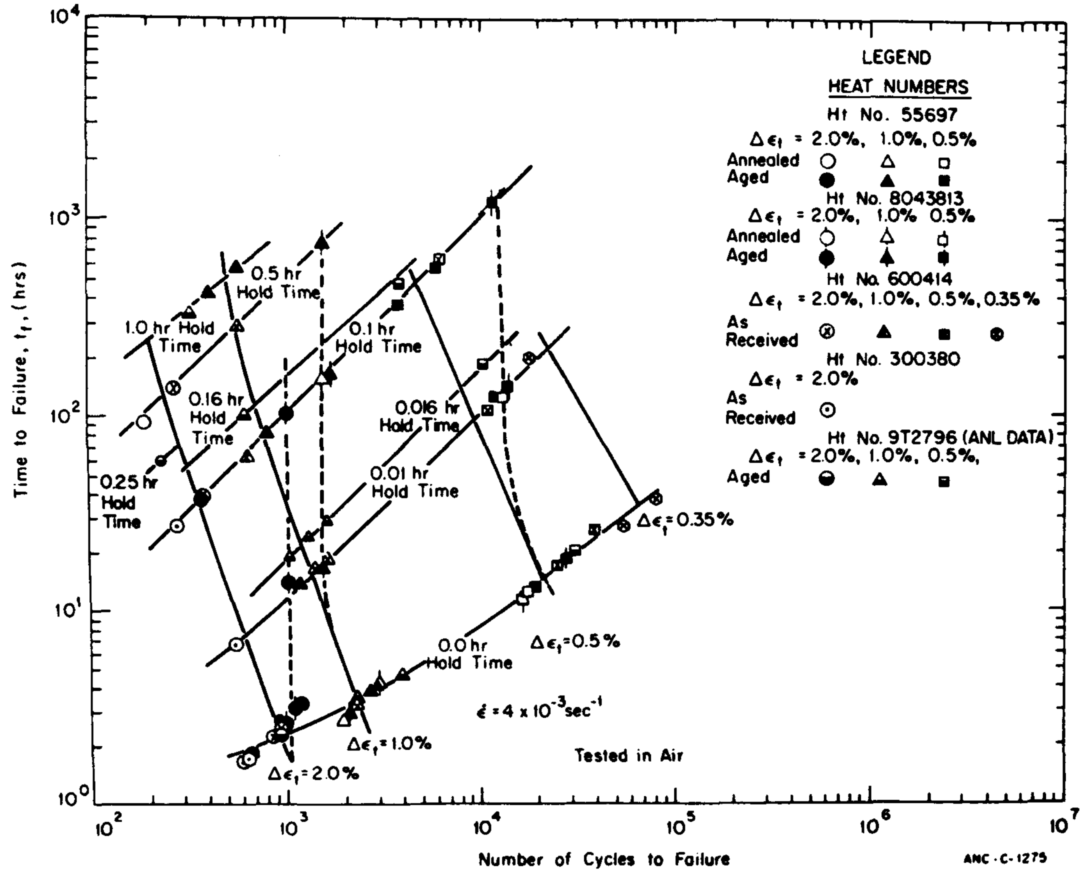


Fig. 23. Time to failure versus number of cycles to failure for five different heats of type 304 stainless steel tested at 593°C (1100°F) with various tensile hold times. *Source:* C. R. Brinkman and G. E. Korth, "Heat to Heat Variations in the Fatigue and Creep-Fatigue Behavior of AISI Type 304 Stainless Steel at 593°C," *J. Nucl. Mater.* 48, 293–306 (1973).

testing condition) than have the other heats. This superiority was attributed to a combination of higher niobium content and thermal mechanical processing history, which resulted in a fine $M_{23}C_6$ precipitate, which strengthened the material. Note in Table 8 that the grain size of this heat is among the smallest and that the stress-rupture lot constant is among the highest, indicating a high stress-rupture strength relative to the other heats. Other investigators have also concluded that a higher uniaxial creep-rupture strength also results in higher creep-fatigue resistance.[47]

The influence of grain size on cycle life of the austenitic stainless steels tested at elevated temperatures with tensile hold times has been studied by several investigators. Data from these studies[48-49] are given in Fig. 24. Although by no means in agreement concerning the magnitude of the effect, these two different sets of data indicate that a reduction in grain size from ASTM 1 to 4 improves cyclic life. However, the extent to which grain size is important may depend on residual element contents and metallurgical state of the material (i.e., the size, shape, and distribution of carbides both within and along the grain boundaries).

Figure 25 shows time-dependent fatigue data for type 316 stainless steel obtained from a single heat of material and tested in the solution-annealed and solution-annealed and aged (before testing) conditions. The data shown are from tests conducted at a single strain range and show that prior thermal aging or thermal aging during testing can markedly increase the creep-fatigue resistance (cycle life) of type 316 stainless steel.[50-51] Figures 26 and 27 show that this same improvement can occur at a lower temperature (i.e., 550°C), at lower strain ranges, and in other heats as well.[52] This improvement in fatigue life has generally been attributed to an increased resistance of the material to grain boundary sliding and cavitation and to the resultant creep crack growth.[53] Similar observations have been noted for type 304 stainless steel.[54]

Table 9 compares the composition and grain size for the three heats of type 316 stainless steel discussed above. Note from Fig. 26 that heat B65808 (abbreviated as 808) has the lowest resistance to creep-fatigue interaction and a grain size (see Table 9) larger than that of heat 83 but not that of heat 297. However, the differences in grain size and in time-dependent fatigue life are not especially important for material in the solution-annealed condition. Table 9 also compares the stress-rupture lot constants [C_h in Eq. (5)] for heats 83 and 297. Heat 83 with the higher lot constant has superior creep-rupture resistance in comparison with lot 297 material as well as superior resistance to creep-fatigue interaction damage as reflected by improved cycle life in the solution-annealed condition, as shown in Fig. 26.

In summary, the above discussion has shown that the metallurgical state of a material (i.e., the presence or absence of intergranular carbides, which can strengthen grain boundaries relative to the matrix

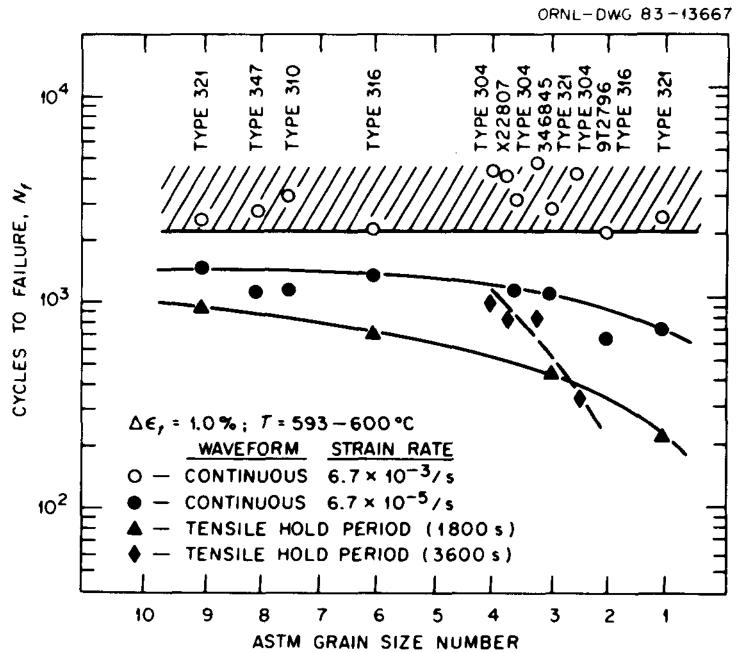


Fig. 24. Variations in grain size result in little or no difference in the low-cycle (continuous) fatigue behavior of the austenitic stainless steels. However, when tensile hold periods are introduced, larger grain sizes reduce fatigue life as shown. Based on data from P. S. Maiya and S. Majumdar, "Elevated Temperature Low-Cycle Fatigue Behavior of Different Heats of Type 304 Stainless Steel," *Metall. Trans. A* 8A, 1651–60 (November 1977) and K. Yamaguchi and K. K. Kanazawa, "Influence of Grain Size on the Low Cycle Fatigue Lives of Austenitic Stainless Steels at High Temperatures," *Metall. Trans. A* 11A, 1691–99 (October 1980).

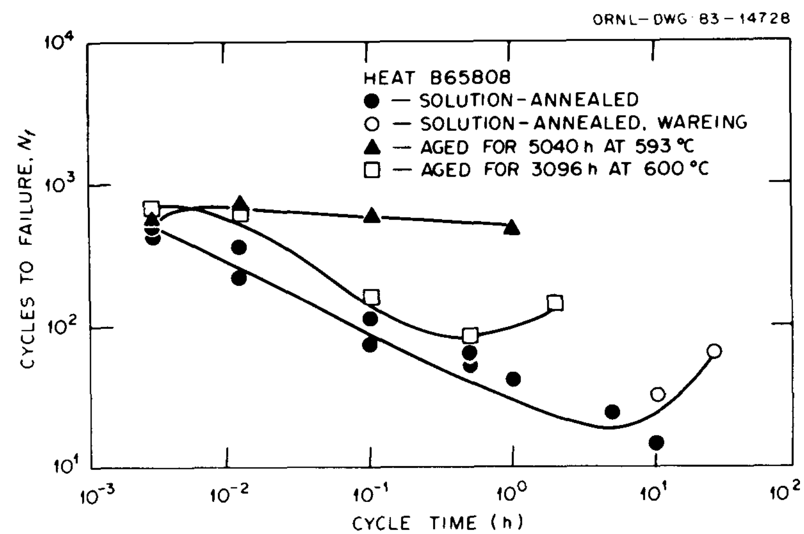
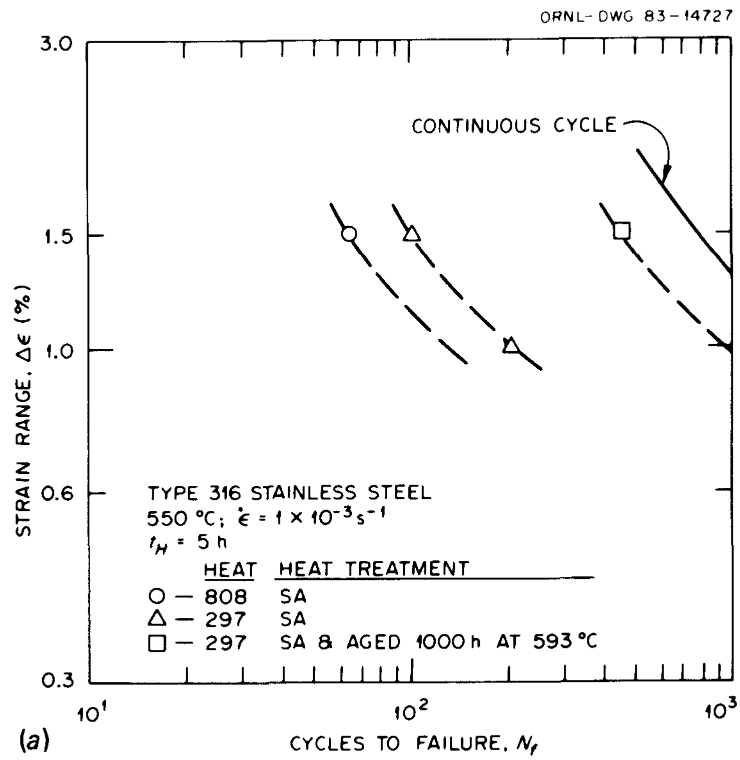


Fig. 25. Increasing the cycle time by increasing the length of tensile hold periods reduces the continuous-cycle fatigue life of type 316 stainless steel tested at 593°C and $\Delta\varepsilon_t$ = 2.0%. However, prior thermal aging or metallurgical changes occurring during prolonged elevated-temperature testing improve fatigue life. Data for two tests courtesy of J. Wareing, UKAEA Springfields Laboratory, Springfields, U.K.



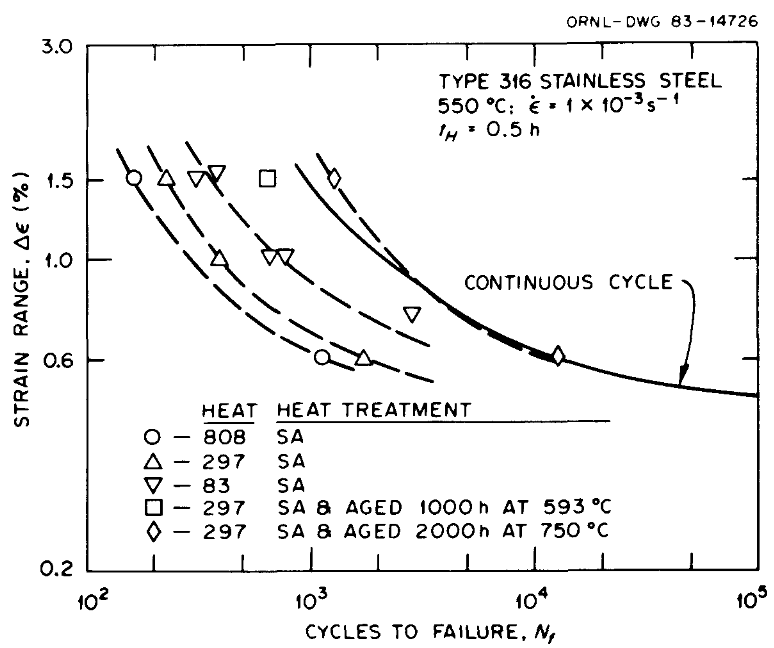


Fig. 26. Comparison of cyclic creep-fatigue data for a tensile hold time of (*a*) 0.5 h and (*b*) 5 h for three heats and two heat treatments. Note that a solution annealing (SA) plus aging heat treatment markedly increases fatigue life relative to material in the solution annealed condition. *Source of data:* D. T. Raske, "Effects of Grain Size on the High-Cycle Fatigue Behavior of Type 316 Stainless Steel," p.1-1–7 in *Mechanical Properties Design Data Program Semiannual Progress Report for Period Ending January 31, 1983*, ORNL/MSP/1.3-83/1, April 1983.

Table 9. Comparison of compositions, grain sizes, and lot constants for three heats of type 316 stainless steel

Heat	Chemical composition (wt %)[a,b]												Product form (mm)	Reannealed grain size		Stress-rupture lot constant
	C	Mn	P	S	Si	Cr	Ni	Mo	Cu	Ti	Co	N		ASTM	(μm)	
83[a]	0.060	1.82	0.038	0.032	0.46	17.11	12.62	2.56	0.37	0.01	0.24	0.044	35, Bar	6.5	33[b]	−11.174
297[a]	0.068	1.90	0.024	0.020	0.64	17.01	13.36	2.49	0.07	<0.01	0.02	0.034	16, Plate	4.3	74[c]	−12.141
808[d]	0.086	1.73	0.010	0.006	0.52	18.16	13.60	2.47	0.078		0.074	0.050	16, Bar	4.9	57[c]	Unknown

[a]*Source of data*: M. K. Booker et al., *A Comparative Analysis of British and American Tensile and Creep Data for Type 316 Stainless Steel*, ORNL/BRP-80/6, June 1980. Lot constants are C_h — in log $t_r = C_h$ 0.1312σ — 2.552 log σ + 20,880/T.

[b]D. T. Raske, "Low-Cycle and Creep-Fatigue Behavior of Type 316 Stainless Steel," pp. 1-1–15, in *Mechanical Properties Design Data Program Semiannual Progress Report for Period Ending Jan. 31, 1981*, ORNL/MSP/1.3-81/1.

[c]D. T. Raske, "The Effect of Metallurgical Variables on the Cyclic Creep-Fatigue Behavior of Type 316 Stainless Steel," pp. 1-1–13, in *Mechanical Properties Design Data Program Semiannual Program Report for Period Ending July 31, 1981*, ORNL/MSP/1.3-81/3.

[d]*Source of data*: J. B. Conway, R. H. Stentz, and J. T. Berling, *Fatigue, Tensile, and Relaxation Behavior of Stainless Steels*, TID-26135, U.S. Atomic Energy Commission, Washington, D.C., 1975.

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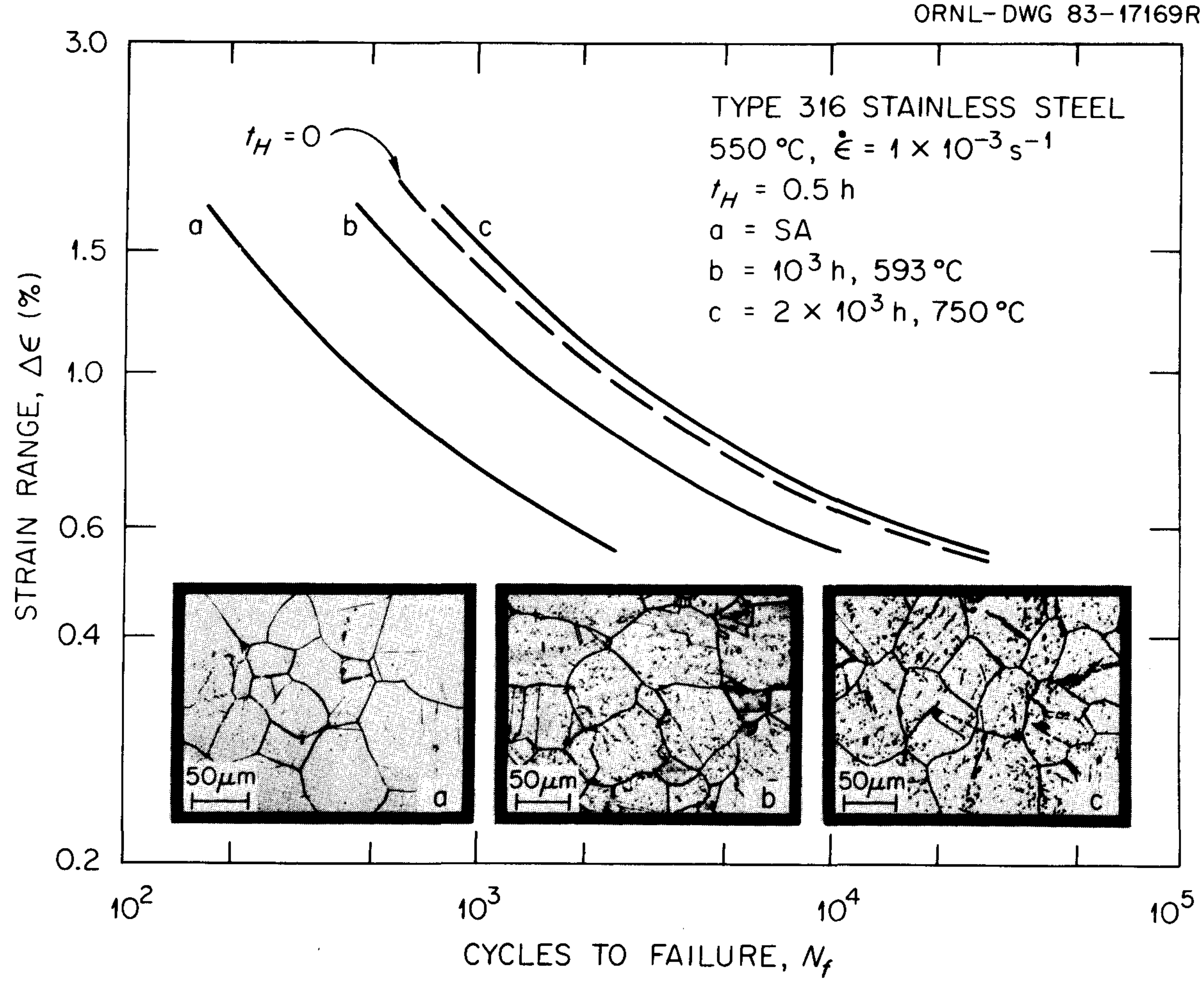

Fig. 27. Comparison of cyclic creep-fatigue data for a tensile hold time of 0.5 h (sensitization) for several heat treatments. Note that a solution-annealing (SA) plus aging heat treatment before testing markedly increases fatigue life relative to material in the solution-annealed condition. Note also precipitation of carbides along grain boundaries. *Source:* J. A. Horak, V. K. Sikka, and D. T. Raske, "Review of Mechanical Properties and Microstructure of Types 304 and 316 Stainless Steel After Long-Term Aging," paper presented at IAEA International Working Group on Fast Reactors Specialists' Meeting on Mechanical Properties of Structural Materials Including Environmental Effects, Chester, England, October 10–14, 1983, to be published.

material) has the most pronounced influence on the time-dependent fatigue behavior of these steels. Residual elements (such as N, Nb, or B) that increase the rupture strength as reflected by the stress-rupture lot constants (Fig. 28) could well improve resistance to creep-fatigue damage (improved cycle life) insofar as their presence improves grain boundary strength relative to that of the matrix.[36-37] However, the presence of both high nitrogen and niobium contents along with preexisting intergranular voidage resulting from irradiation damage or some other source may be particularly damaging in reducing fatigue life.[55] This can occur if the matrix is strengthened relative to the grain boundaries. Minimizing the grain size may improve time-dependent fatigue life, particularly in the range of ASTM grain sizes 1 to 4. However, the influence of grain size on long-term creep-fatigue behavior is still unknown and requires further clarification. Finally, some evidence suggests that the creep-rupture lot constant concept may be used as an index of a material's resistance to creep-fatigue damage. Thus, the same steps taken to improve creep rupture strength and to decrease scatter may also be expected to have a similar influence on time-dependent fatigue behavior.

BILINEAR STRESS-STRAIN HARDENING

The Nuclear Regulatory Commission standards (e.g., RDTF9-5T, March 1981) recommend bilinear representation of both the monotonic and cyclic stress-strain response in elastic-plastic analysis. This procedure requires the use of a number of hardening parameters, which depend on temperature and possibly on strain rate and are defined for monotonic loading as follows:

E = Young's modulus,

E_p = plastic modulus,

σ_0 = bilinear yield point, defined by intersection of two lines whose slopes are E and E_p,

κ_0 = ${\sigma_0}^2/3$, and

C = $2EE_p/3(E - E_p)$.

From Eqs. (1) and (2), one would expect that variations in composition (e.g., carbon and nitrogen contents) as well as in grain size, which

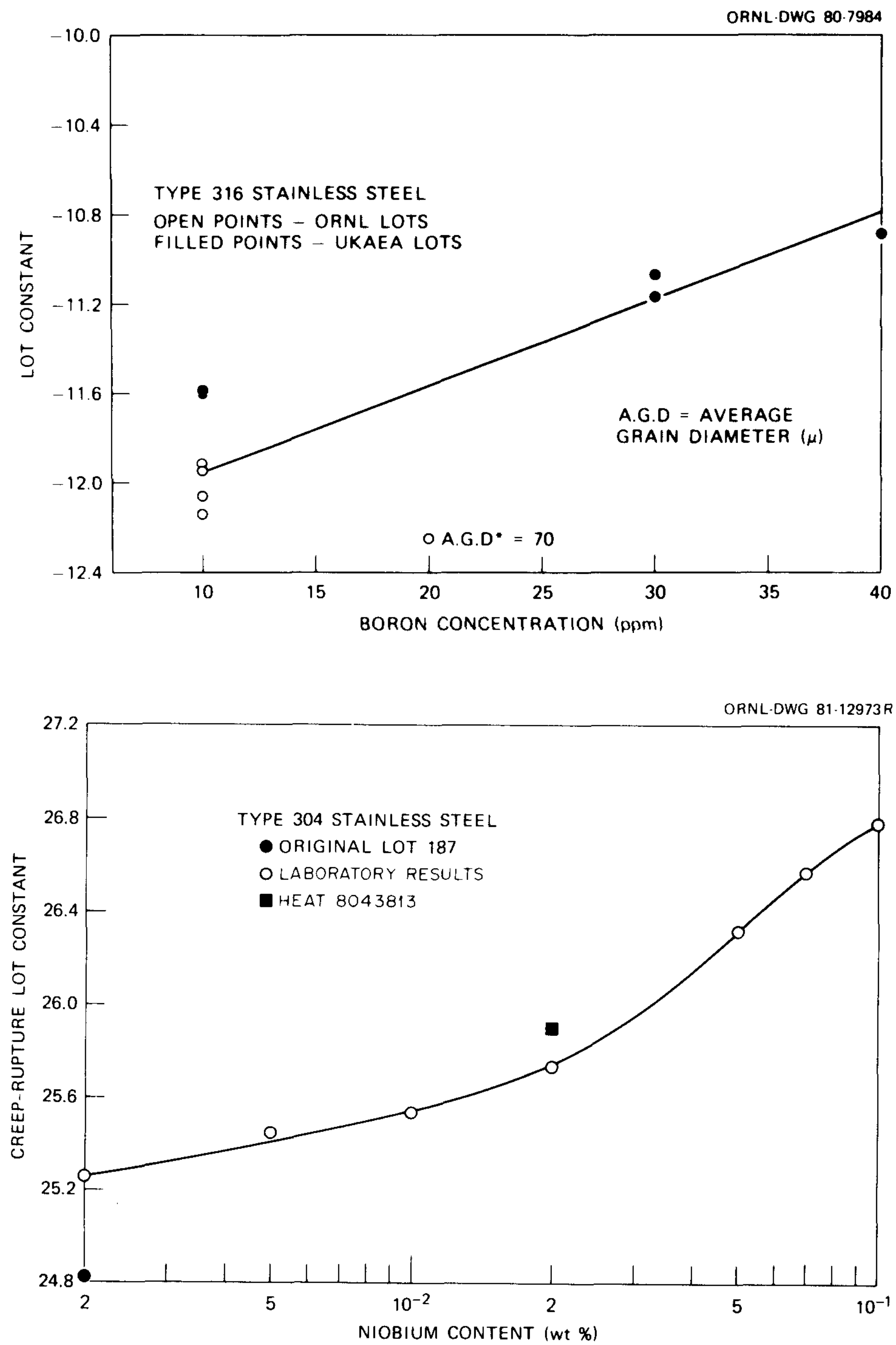

Fig. 28. Small changes in the boron and niobium contents have a marked effect on the stress-rupture lot constants.

influence the yield and ultimate strengths, would similarly influence σ_0 and E_p and therefore, σ_0 and C. Figure 29 shows the initial portions of the monotonic tensile stress-strain curves of two heats of type 304 stainless steel tested at 593°C. They are compared with the hot tensile curve taken from the ASME Code. The lower yield behavior of these two heats in comparison with the code curve is probably because they were tested after reannealing, which removed any residual cold work. Heat 796K with the lower carbon content and substantially larger grain size shows a significantly reduced initial bilinearized yield (i.e., 62 vs 93 MPa) in comparison with heat 330. Similarly, E_p, κ_0, and C values are also reduced. We conclude, from the limited results of calculations given in Fig. 29, that significant variations in variables such as grain size, residual cold work, thermal history, and carbon and nitrogen contents are likely to result in marked variations in the monotonic bilinear parameters.

Bilinear representation is also required for elastic-plastic loadings after the initial loading. These parameters are obtained from cyclic stress-strain curves or hysteresis loops generated in strain control. Most of the work reported to date aimed at defining the cyclic constants for type 304 stainless steel was performed on a single heat of type 304

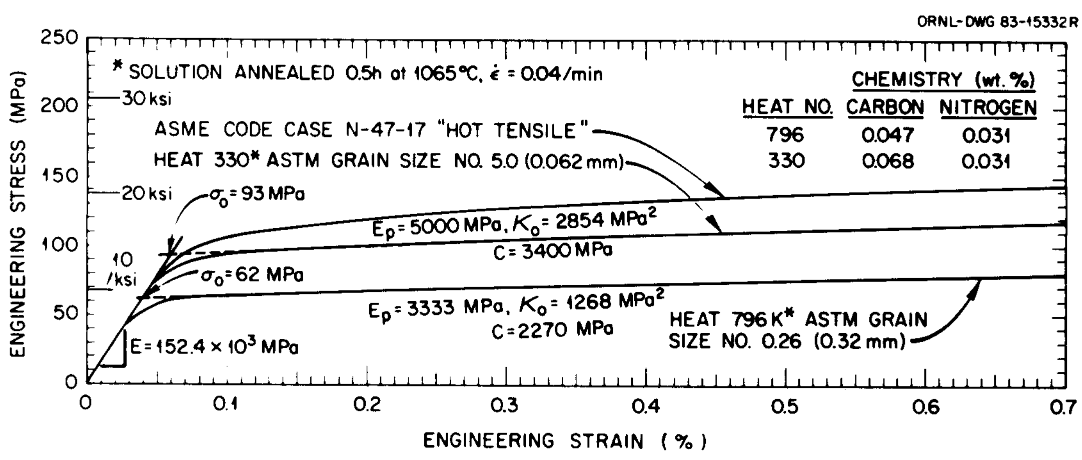

Fig. 29. Comparison of the initial portion of the monotonic stress-strain curves of two heats of type 304 stainless steel with a code case hot tensile curve obtained at 593°C. An increase in grain size results in a decrease in yield strength with resultant changes in the bilinear hardening parameters. Bilinear parameters σ_0, E_p, C, and κ_0 were calculated for a maximum strain of 0.3%.

stainless steel.[46,56-57] Heat-to-heat variations are known, however, and grain size, thermal aging, and minor changes in contents of certain key elements are expected to influence the magnitude of the bilinear constants. The fact that grain size is an important variable is again seen in Fig. 30, where first-cycle stress range is plotted as a function of grain size in a Hall-Petch form for a number of cyclic tests conducted on a single heat of type 316 stainless steel.[58] Note that, as the grain size decreases, the stress range required to produce a controlled strain range increases markedly. Because this strain range is of the magnitude found in design situations, grain size variations need to be considered in specifying appropriate hardening parameters, at least during the first few cycles.

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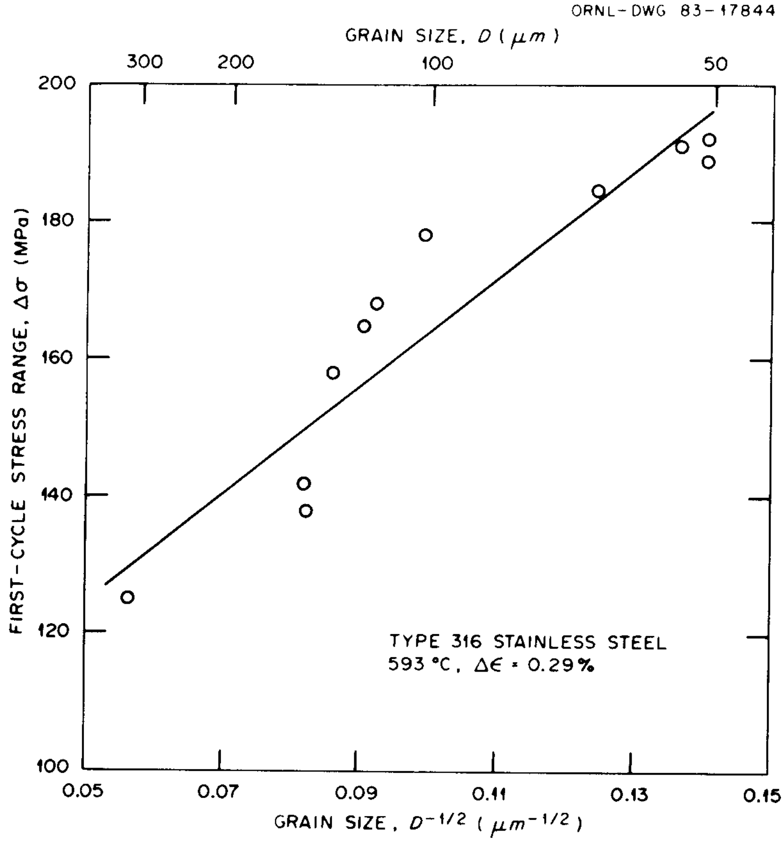

Fig. 30. First-cycle stress range as a function of grain size for solution-annealed type 316 stainless steel. *Source of data*: D. T. Raske, Argonne National Laboratory.

In conclusion, heat treatment, chemical composition, and particularly grain size differences can change the initial monotonic yield point and other parameters used to define the bilinear stress-strain curve. These same variables are also likely to modify material response, particularly in the first few cycles under cyclic strain-controlled loading conditions while the material is undergoing significant changes in hardening.

FATIGUE CRACK PROPAGATION

Fatigue crack growth in the austenitic stainless steels has been studied by many investigators. Generally, they have concluded that, under conditions of continuous cycling, crack propagation rates do not depend on heat-to-heat variations and grain size for conditions leading to transgranular crack propagation in the intermediate growth rate range (i.e., in the range where the Paris law or power law between crack growth rate and stress intensity is applicable).[59-60] However, two situations at high temperature need to be considered. The first is crack growth behavior in the low-stress-intensity (ΔK) or threshold region, and the second is at high ΔK levels and at low frequencies under tensile creep loading conditions, where accelerated crack growth may occur. These two possibilities are now briefly discussed in terms of metallurgical implications.

At low ΔK levels (i.e., below about 25 $MPa \cdot \sqrt{m}$) or in the near-threshold regime, where environmental factors are particularly important, there is some question about the effects of metallurgical variables such as grain size on crack growth rate behavior. For example, decreasing the grain size in low-strength steels [i.e., yield strengths less than 500 MPa (73 ksi)], reduces the threshold stress intensity value.[61] This is just the opposite of what one normally sees when viewing strain-range fatigue-life curves (see Fig. 22), where a decrease in grain size increases the yield strength and the high-cycle fatigue life by an increase in resistance to crack initiation rather than by crack arrest. However, for high-strength steels [i.e., yield strength values greater than 1300 MPa (190 ksi)], grain size differences have been shown to have no effect on threshold values.[61] Other investigators, however, have reported that the threshold stress intensity is nearly independent of all material properties except the elastic modulus.[62] French investigators have reported that, in the case of type 316 stainless steel, no true threshold exists (at least down to growth rates of 10^{-7} mm/cycle).[63] The influence of grain size on near-threshold crack growth behavior in type 316 stainless steel needs further investigation at elevated temperatures in support of applications involving thermal striping. For example, judging from the strain-life curve given in Fig. 22, one might specify a small-grain material to take advantage of increased resistance to crack initiation. However, preexisting flaws present as a consequence of fabrication may actually grow at somewhat higher rates because of the reduced grain sizes. Any additional experimental work undertaken, however, should be conducted in a way to control environmental parameters because of their known influence on crack propagation in the near-threshold region.

The second region in which metallurgical variables are likely to be important in the crack growth rate versus ΔK curve is at fairly high stress intensities (i.e., above 30 $MPa \cdot \sqrt{m}$) at which accelerated crack propagation can occur (see Fig. 31) (ref. 64). Accelerated crack propagation occurs at temperatures and loading conditions that lead to intergranular crack propagation.[65–66] Like creep-fatigue interaction that occurs under strain-controlled conditions, accelerated crack propagation is a function of metallurgical history[67] and is likely to depend strongly on thermal mechanical processing history.[68] Material variables that are thought to be important are those that lead to grain boundaries containing defects[55] and any other variables [such as high nitrogen content[29] (Fig. 12), or cold work] that can lead to reduced creep-rupture ductility. A material with below normal creep-rupture ductility would be expected to show an increased tendency to display accelerated crack growth behavior under tensile loading conditions and at temperatures at which creep-induced grain boundary cavitation can occur.

In summary, material variables such as heat-to-heat variations in composition, grain size, and small levels of cold work do not influence fatigue crack propagation in types 304 and 316 stainless steel. The two exceptions to this conclusion are at low stress intensities or in the

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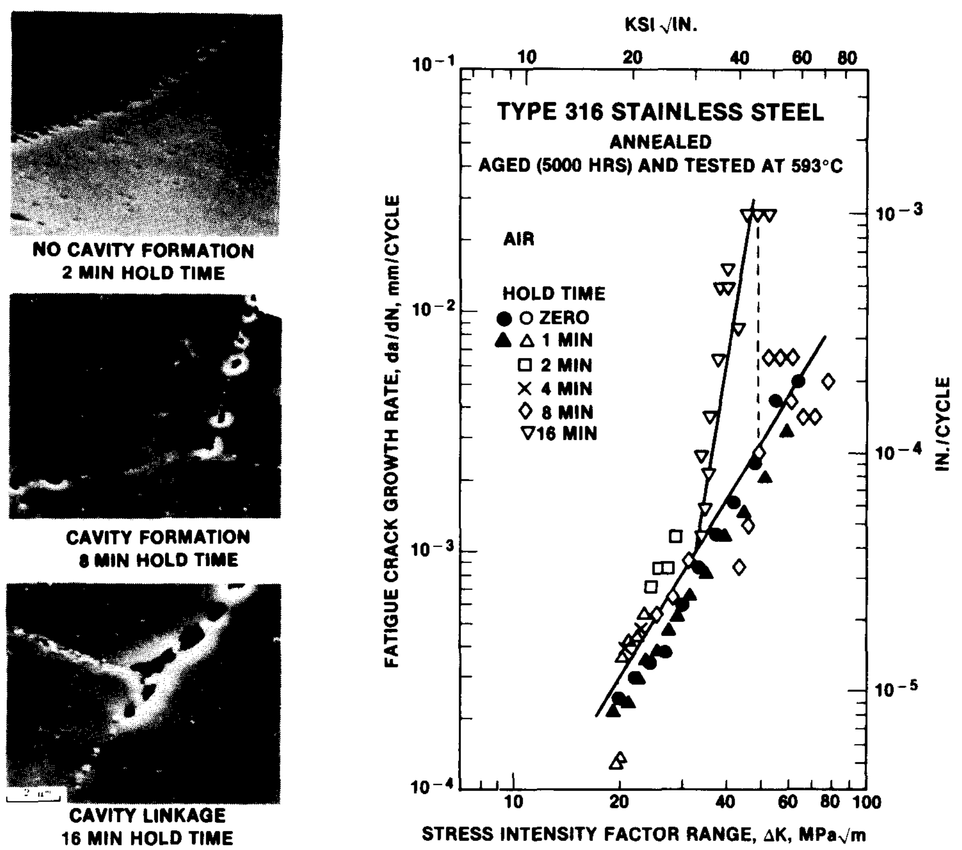

Fig. 31. Tensile hold periods longer than 960 s (16 min) can cause accelerated crack propagation in type 316 stainless steel tested at 593°C by cavity linkage at grain boundaries. Figure courtesy of D. J. Michel, Naval Research Laboratory, Washington, D.C.

near-threshold region, where grain size variations may be important at high temperatures (i.e., 500°C and above, depending on the environment), and high-stress-intensity tensile creep loading conditions under which material variables that induce grain boundary cavitation are likely to show significant effects.

TOUGHNESS

Brittle or unstable fracture is normally not a design consideration in the austenitic stainless steels slated for use in Liquid Metal Fast Breeder Reactor (LMFBR) systems. Their face-centered cubic structure prevents them from showing a true ductile-to-brittle transition temperature, and thus they inherently resist brittle fracture. Thermal aging at elevated temperatures does, however, change the toughness properties when measured at room temperature, as shown in Fig. 32 for type 316 stainless steel.[69] Impact energy decreases continuously with increasing exposure time. Central Electricity Generating Board (CEGB) investigators have reported drops in impact energy from 299 to 88 J for type 316 stainless steel after service exposure to 8.4×10^4 h at 570°C (ref. 70). Exposure of type 316L steel (~0.03% C) produced only small changes in toughness values. The reduction in room-temperature toughness is attributed to the presence of carbides and sigma phase. Hence, carbon and initial ferrite levels as well as method of testing appear to be important material variables.

Mills[71–74] has spent considerable effort in studying heat-to-heat variability in the austenitic stainless steels and their weldments by

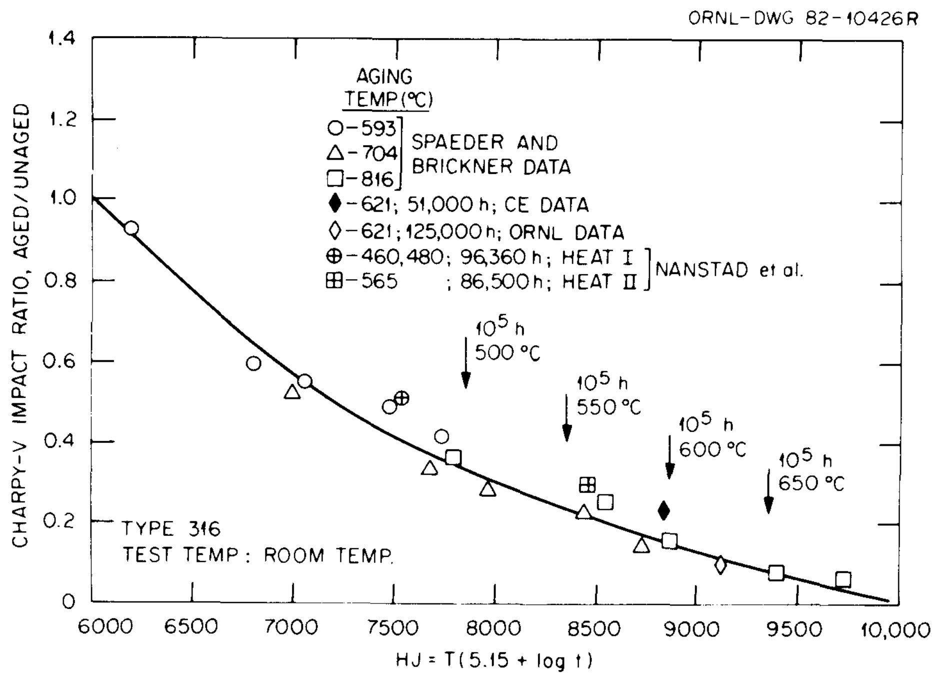

Fig. 32. Charpy V-notch impact energy ratio (aged-to-unaged) at room temperature as a function of the Holloman-Jaffe parameter for several heats of type 316 stainless steel tested at ORNL and elsewhere. *Source:* V. K. Sikka, *Effects of Thermal Aging on the Mechanical Properties of Type 316 Stainless Steel — Elevated Temperature Properties*, ORNL/TM-8371, October 1982.

fracture mechanics techniques. Some of his work in the form of R-curves[72] is shown for several heats of type 304 stainless steel, tested at 427 and 538°C, in Fig. 33. Note that the toughness values J_{Ic} are high but that heat-to-heat variability is considerable. Heat E shows the lowest toughness values, but its composition is comparable to that of the other heats.[72] The yield strength was somewhat higher and the uniform elongation lower in heat E than is the other heats. Other than these subtle variations, no real differences were apparent in other properties, including grain size, which varied from about ASTM 2 to 4 in these heats. The reasons for the differences in toughness between these heats probably lie in subtle differences in the microstructure, such as carbide or inclusion density, morphology, and distribution. Indeed, Mills[73] has noted that specimen orientation is an important variable in stainless steel

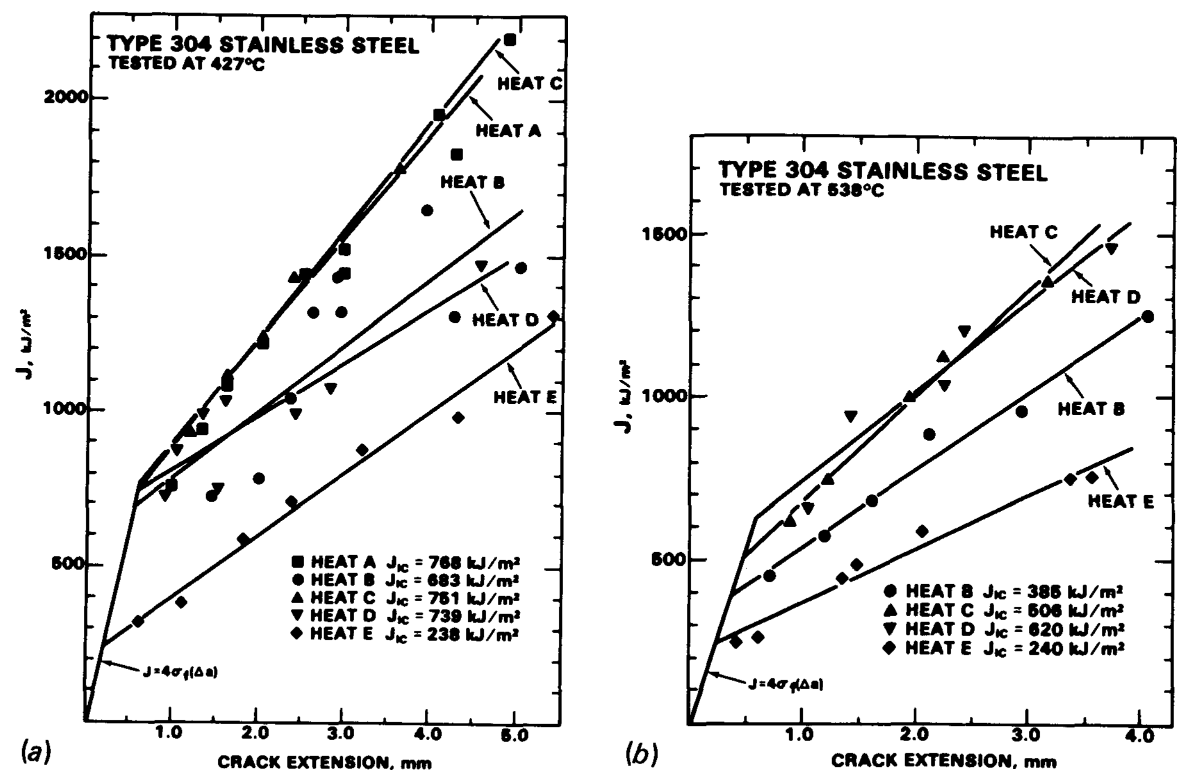

Fig. 33. Toughness R-curves for type 304 stainless steel. (*a*) 427°C. (*b*) 538°C. *Source*: W. J. Mills, "Survey of Heat-to-Heat Variations on the Fracture Toughness of Type 304 Stainless Steel," pp. 3-8–13 in *Mechanical Properties Design Data Program Semiannu. Prog. Rep. for Period Ending Jan. 31, 1981*, ORNL/MSP/1.3-81/1, 1981.

piping material because of the presence of stringers of impurities. Lower toughness was found when the crack propagation was parallel to the stringers or in the axial direction of the pipe. This suggests that melting practice may be an important variable, with electroslag remelting (ESR) and vacuum arc remelting practices producing better properties because the material is cleaner.

The importance of that microstructure can be seen in Fig. 34, where R-curves are compared for several stainless weld metals deposited by various processes.[74] Note that the gas tungsten arc (GTA) process, which produces a finer microstructure than does the submerged arc (SA) welding process, produces the higher toughness.

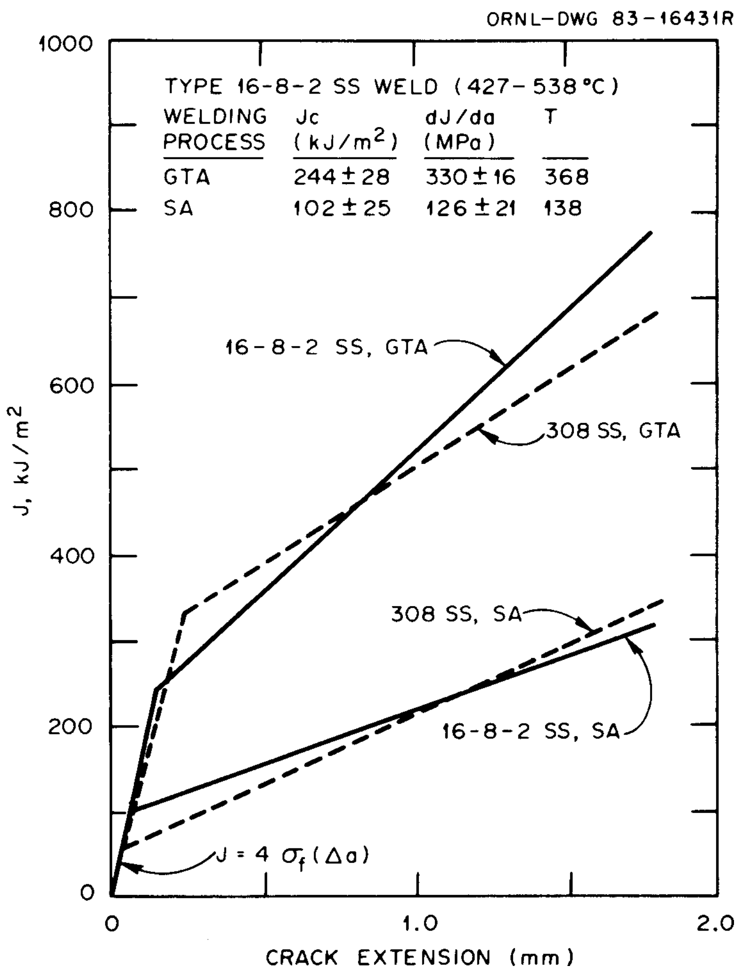

Fig. 34. Comparison of R-curves for 16-8-2 and type 308 stainless steel weld metal deposited by several different processes.

In summary, the austenitic stainless steels and their weldments generally have adequate toughness to prevent failure by stable or unstable fracture at stress levels up to and including yield values for small and medium-size flaws. Thermal aging reduces the toughness somewhat, depending on the type of test conducted (Charpy impact vs R-curve technique). Microstructural differences introduced by material processing, carbon content, weld deposition method, and possibly grain size all contribute to heat-to-heat variability. Additional work is needed on forging material, in which large variations in grain size are possible.

DISCUSSION

As described in the introduction, both French and British LMFBR designers have attempted to optimize the chemical composition of type 316 stainless steel. Both the French and British specifications along with actual heats produced to or close to these specifications are compared with U.S. ASME specifications in Table 10. The French steel is actually a type 316L grade with higher nitrogen and slightly higher boron contents than those of the U.S. type 316 stainless steel, which typically contains 0.010 to 0.07 wt % N and 0.0001 to 0.0005 wt % B. Similarly, the British also specify nitrogen and boron contents. As was previously shown, these two elements are strengtheners and compensate for reduced carbon levels, particularly in the French steel. Both the British and French also restrict the Si, S, and P contents and narrow the range on Mn, Ni, Cr, and Mo contents relative to U.S. type 316 stainless steel. Narrowing the range of some of these elements is likely to reduce data scatter in long-term creep-related properties, as was shown to be the case for tensile properties in the introduction of this report. The higher boron content increases the rupture lot constant value as shown in Fig. 28, with resultant increased rupture strength (Fig. 35). The increased rupture strength is thought to be due to the formation of grain boundary precipitates [e.g., Cr_2B and $M_{23}(C,B)_6$], which strengthen the grain boundaries, thereby inhibiting grain boundary cavitation.[70] This would also improve resistance to creep-fatigue damage by impeding intergranular crack propagation. As

Table 10. Comparison of British and French chemical specifications for type 316 stainless steel with American Society for Mechanical Engineers requirements

Element	French specification	Heat 1	British specification	Heat 83[a]	ASME specification
Carbon	<0.03	0.023	0.03–0.06	0.06	0.04–0.08
Nitrogen	0.06–0.08	0.076	0.04–0.07	0.04	
Boron	0.0015–0.0035	0.0030	0.002–0.005	0.0017	
Silicon	<0.50	0.40	0.2–0.6	0.49	<1.0
Manganese	1.6–2.0	1.70	1.6–2.0	1.75	≤2.0
Nickel	12.0–12.50	12.12	11.0–12.5	12.30	10.0–14.0
Chromium	17.0–18.0	17.44	16.5–18.0	17.68	16.0–18.0
Molybdenum	2.3–2.7	2.45	2.0–2.75	2.34	2.0–3.0
Sulfur	<0.025	0.006	<0.015	0.021	<0.03
Phosphorus	<0.035	0.029	<0.03	0.035	≤0.045
Cobalt		0.23	<0.15		
Copper		0.17	<0.25		
Titanium		<0.01	<0.02		
Niobium		<0.01	<0.02		

[a]This heat is slightly off the specified boron and sulfur contents and was produced before formulation of the final specification. It is a well-characterized heat at Risley Nuclear Laboratories.

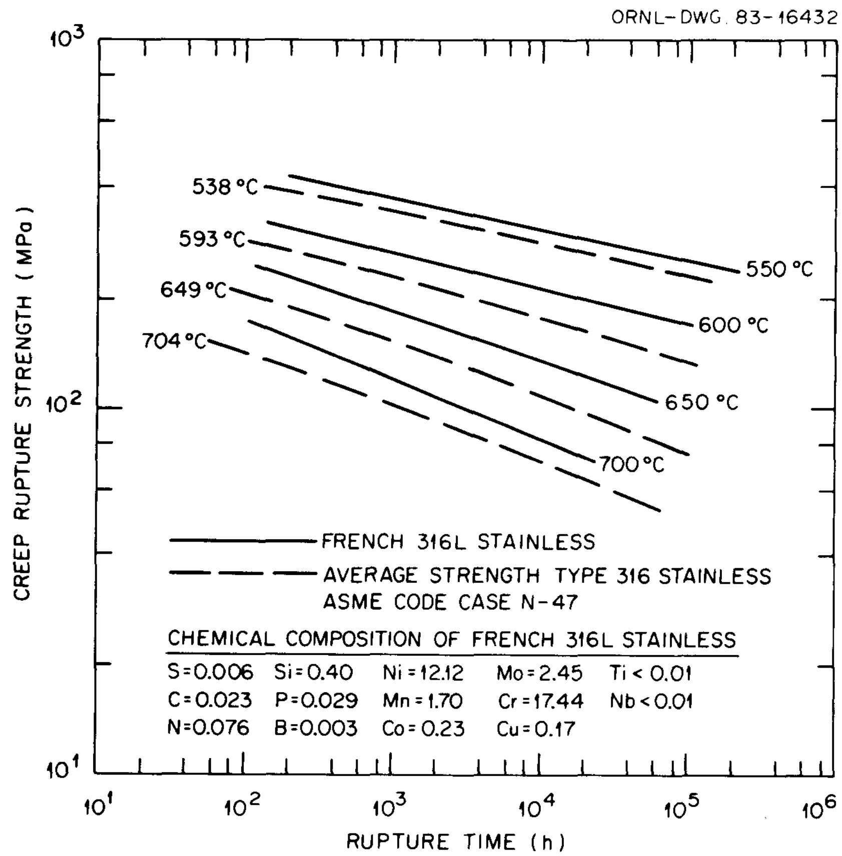

Fig. 35. Comparison of creep-rupture strength of French type 316L stainless steel with average strength of type 316 stainless as defined by ASME Code Case N-47. Note increased strength of French steel used in LMFBRs. *Source of French data*: M. Mottot et al., "Behavior in Fatigue-Relaxation of a High-Creep Resistant Type 316L Stainless Steel," pp. 152–68 in *Low-Cycle Fatigue and Life Prediction*, ASTM STP 770, American Society for Testing and Materials, Philadelphia, 1982.

discussed previously, however, concentrations of boron at grain boundaries as a consequence of slow cooling or prolonged thermal exposure could creep embrittle the material because of the presence of helium following exposure to a thermal neutron fluence. Only about 10^{-4} at. ppm He is necessary for the onset of this type of embrittlement.[75] Large grain sizes are also undesirable where this type of embrittlement is possible.[75]

In reactor core applications, the potential deleterious effect of boron has been recognized by U.S. radiation effects experts and by limitations set on both grain size and boron content, as shown in Table 3.

In applications to which damage from thermal neutron irradiation effects is not important but which require greater creep strength, addition of boron up to 15 to 35 ppm is desirable.

The intent of this report was to recommend more restrictive composition, melting, and fabrication practice within existing specifications for the austenitic stainless steels to reduce data scatter, particularly in high-temperature creep-related properties. These changes should not significantly modify the strengths. In considering most of the mechanical properties in this report of interest to the designer, grain size variations have been shown to be a particularly significant cause of heat-to-heat variability. Generally, a small grain size is most desirable to optimize behavior; however, it may be impractical in product forms such as forgings to specify a grain size range. The best cost-effective measure that can be taken with present understanding of the long-term elevated-temperature behavior of these steels is to recommend specifications that serve as target objectives, that involve "good practice," and that reduce the range of allowable chemical composition variables on those elements with the greatest influence, particularly on creep-related properties. These recommendations in comparison with current ASTM and ASME specifications are listed in Table 11, and the rationale for these recommendations follows.

To reduce strength variations, the carbon spread is decreased from 0.04 to 0.02%, and the nitrogen spread is controlled at 0.03%. The carbon level is reduced slightly and the nitrogen content is increased slightly to improve long-term creep strength and possibly ductility[76] and creep-fatigue resistance. An upper limit is placed on nitrogen because it markedly increases the yield and work hardening rate and thereby reduces formability. Decreasing carbon content slightly may increase creep-rupture ductility.[76]

Manganese is an austenite former with little or no effect on short-term strength-related properties. However, 1 to 2 wt % Mn may improve long-term creep-rupture-strength properties[29] at temperatures near 600°C (ref. 7) [see Eq. (8)] and rupture ductility.[76]

Table 11. Recommended composition limits for stainless steels compared with current specifications by ASTM and ASME

Element	Content in type 304 (wt %)		Content in type 316 (wt %)	
	Current	Recommended	Current	Recommended
Carbon	0.04–0.08	0.04–0.06	0.04–0.08	0.04–0.06
Nitrogen		0.04–0.07		0.04–0.07
Silicon	<1.0	<0.6	<1.0	<0.6
Manganese	<2.0	1.0–2.0	<2.0	1.0–2.0
Nickel	8.00–10.50	8.00–10.00	10.0–14.0	11.00–12.5
Chromium	18.00–20.00	18.5–20.00	16.0–18.0	17.00–18.00
Molybdenum		<0.2	2.0–3.0	2.5–3.0
Sulfur	<0.03	<0.02[a]	<0.03	<0.02[a]
Phosphorus	<0.045	<0.045[a]	<0.045	<0.03[a]
Niobium		<0.02[a]		
		(ppm)		(ppm)
Aluminum		500		500
Antimony		200		200
Boron				30[b]
Lead		30		30
Selenium		150		150
Tin		150		150
Vanadium		500		500
Zinc		100		100

[a]To reduce data scatter further specify a minimum value of 0.005%.

[b]To reduce data scatter further specify a minimum value of 15 ppm.

Silicon is a mild ferrite former whose presence can increase both the tensile yield and ultimate strengths [see Eqs. (1) and (2)]. That it can reduce creep rupture strength in type 316 stainless steel [Eqs. (8) and (9)] has been reported.[7]

Nickel is a strong austenite former. To increase the probability of a small amount of ferrite being present, the allowable range must be reduced slightly. Variations have little effect on short-term strength properties. Reducing the nickel content slightly may be beneficial to the long-term stress-rupture properties [Eq. (8)].

Molybdenum content is increased slightly within the present allowable range in type 316 stainless steel to maintain strength and reduce strength variations. This element is a ferrite former.

Chromium is a ferrite former, and variations exert only a small influence on short-term strength properties. The chromium content is targeted to the upper end of the current allowable range to increase the probability of some ferrite.

Boron, niobium, and titanium can have a marked influence on the elevated-temperature properties by several mechanisms, including helping to maintain a small grain size. From 1 to 10 ppm B increases creep ductility. Beyond this range (Fig. 28), it markedly increases the rupture strength of type 316 stainless steel. Similarly, small variations in niobium content in the range from 50 to 1000 ppm (Fig. 28) can have a marked influence on rupture strength of type 304 stainless steel. If improved creep properties are desired, these elements can be targeted to higher levels; otherwise, their contents should be restricted to a narrow range if possible.

Other steps during material procurement are recommended as good practice and should be followed to minimize variations in properties while optimizing behavior.

1. Recommend a melting practice; AOD or combination of AOD and ESR, because this will improve quality (cleanliness and less directionality in properties of the steel).
2. Request a given pouring practice; bottom versus top (bottom is preferred).

3. Recommend a fabrication process; for example, a billet can be hot forged (coarse grain) or hot rolled (fine grain).
4. Specify grain size in the range of ASTM 3 through 6 (40–102 μm) whenever practical.
5. Control levels of cold or warm work present whenever possible.

Finally, note that not all these recommendations are necessarily appropriate to long-term service at temperatures above 649°C.

CONCLUSIONS

Variations in specific mechanical properties of types 304 and 316 stainless steel were investigated, and the following conclusions were made.

1. Variations in tensile properties of these steels are due to the current specifications, which permit significant variations in C, N, Nb, and B contents, and to the thermomechanical processing history, which results is variations in grain size and levels of cold work. Only limited data are available for forgings, and more data are required to fully define data scatter in this product form.

2. Many of the same variables that impact short-term ultimate tensile strength are also important in determining the creep-rupture and minimum creep rate response of type 304 stainless steel. These variables include material composition and grain size. Hence, grain size control and some limitations on variability in contents of such elements as Nb, Ti, Mo, B, C, and N, when practical, should be considered to minimize data scatter. The current success in predicting both the creep and creep-rupture response of both types 304 and 316 stainless steel with lot-centered analysis techniques suggests the value of a short-term creep test as a means of estimating the appropriate lot constants when calculations of creep deformation are required.

3. In estimating the creep and creep-rupture response of type 316 stainless steel, the relationship to short-term tensile strength is not well defined. Available data from the literature suggest that, in the range 593 to 625°C, control of carbon content to about 0.05% and of other elements, such as B, N, and perhaps S, can reduce data scatter.

4. Currently available data suggest that small levels of cold work normally associated with certain product forms of mill-annealed material make only a small contribution (if any) to the long-term creep response of type 304 stainless steel at temperatures around 593°C. At higher levels of cold work (e.g., 10–30%), significant strengthening will occur in the near term; however, if the temperatures are high enough, recrystallization with resultant decreased rupture life may occur.

5. Control of material composition specifications at the residual element level in both steels is confused by the fact that certain combinations of elements may be strengthening or weakening on prolonged creep at elevated temperatures.

6. It can be concluded from continuous fully reversed fatigue tests that, at strain ranges of 0.5 or less (the range of most interest to the designer), data scatter is due to grain size differences, interlaboratory variations in test techniques, and metallurgical instability of these steels at high temperature. To increase high-cycle fatigue life, the data indicate that grain size should be minimized. Thermal aging may decrease slightly the continuous-cycle fatigue life at lower temperatures by carbide cracking.

7. The metallurgical state of a material (i.e., the presence or absence of intergranular carbides, which can strengthen grain boundaries relative to the matrix material), has the most pronounced influence on the time-dependent fatigue behavior of these steels. Residual elements (such as N, Nb, or B), which increase the rupture strength as reflected by the stress-rupture lot constants, could well improve resistance to creep-fatigue damage (improve cycle life) insofar as their presence improves grain boundary strength relative to that of the matrix. Minimizing the grain size may improve time-dependent fatigue life, particularly in the range of ASTM grain sizes 1 to 4. However, the influence of grain size on long-term creep-fatigue behavior is still unknown and requires further clarification. Some evidence suggests that the creep-rupture lot constant may be used as an index of a material's resistance to creep-fatigue damage. Thus, the same steps taken to improve creep rupture strength and ductility and to decrease scatter may also be expected to have a similar influence on time-dependent fatigue behavior.

8. Differences in heat treatment, composition, and particularly grain size can change the initial monotonic yield point and other parameters used to define the bilinear stress-strain curve. These same variables are also likely to modify material response, particularly in the first few cycles under cyclic strain-controlled loading conditions while the material is undergoing significant changes in hardening.

9. Material variables such as heat-to-heat variations in composition, grain size, and small levels of cold work do not influence fatigue crack propagation in these steels. Two exceptions to this conclusion are at low stress intensities or in the near-threshold region, where grain size variations may be important at high temperatures (e.g., 500°C and beyond, depending on the environment), and under high-stress-intensity tensile creep-loading conditions, under which material variables that induce grain boundary cavitation are likely to show significant effects.

10. Generally the austenitic stainless steels and their weldments have adequate toughness to prevent failure by stable or unstable fracture at stress levels up to and including yield values for small and medium-size flaws. Thermal aging reduces the toughness somewhat, depending on the type of test conducted (Charpy impact vs R-curve technique). Microstructural differences introduced by material processing, carbon content, weld deposition method, and possibly grain size all contribute to heat-to-heat variability. Additional work needs to be conducted on forged material, in which large variations in grain size are possible.

11. Specific recommendations are made concerning a slightly more restrictive composition specification aimed at reducing heat-to-heat variations, particularly in creep properties. Recommendations are also given and defined as "good practice" to be followed whenever practicable, such as melting practice and grain size restrictions. We feel that these recommendations are appropriate for material to be subject to long-term service at temperatures below 649°C.

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Appendix

LOT CONSTANT VARIATIONS WITH GRAIN SIZE AND ELEMENT CONTENTS

Figures A.1 and A.2 give the creep-rupture lot constants as functions of grain size and composition for types 304 and 316 stainless steel, respectively. Each data point shown is the lot constant for a given heat with the indicated composition or grain size. The intent here was simply to show major trends in rupture strength. A numerical increase in the lot constant indicates increased rupture strength.

These constants are the C_h, which reflect the heat-to-heat variation, in the equations for rupture strength:

Type 304 stainless steel:

$$\log t_r = C_h - 0.01258T - 1.232 \times 10^{-5}T\sigma - 0.004770T \log \sigma \ . \qquad \text{(A.1)}$$

Type 316 stainless steel:

$$\log t_r = C_h - 0.01312\sigma - 2.552 \log \sigma + 20{,}880/T \ . \qquad \text{(A.2)}$$

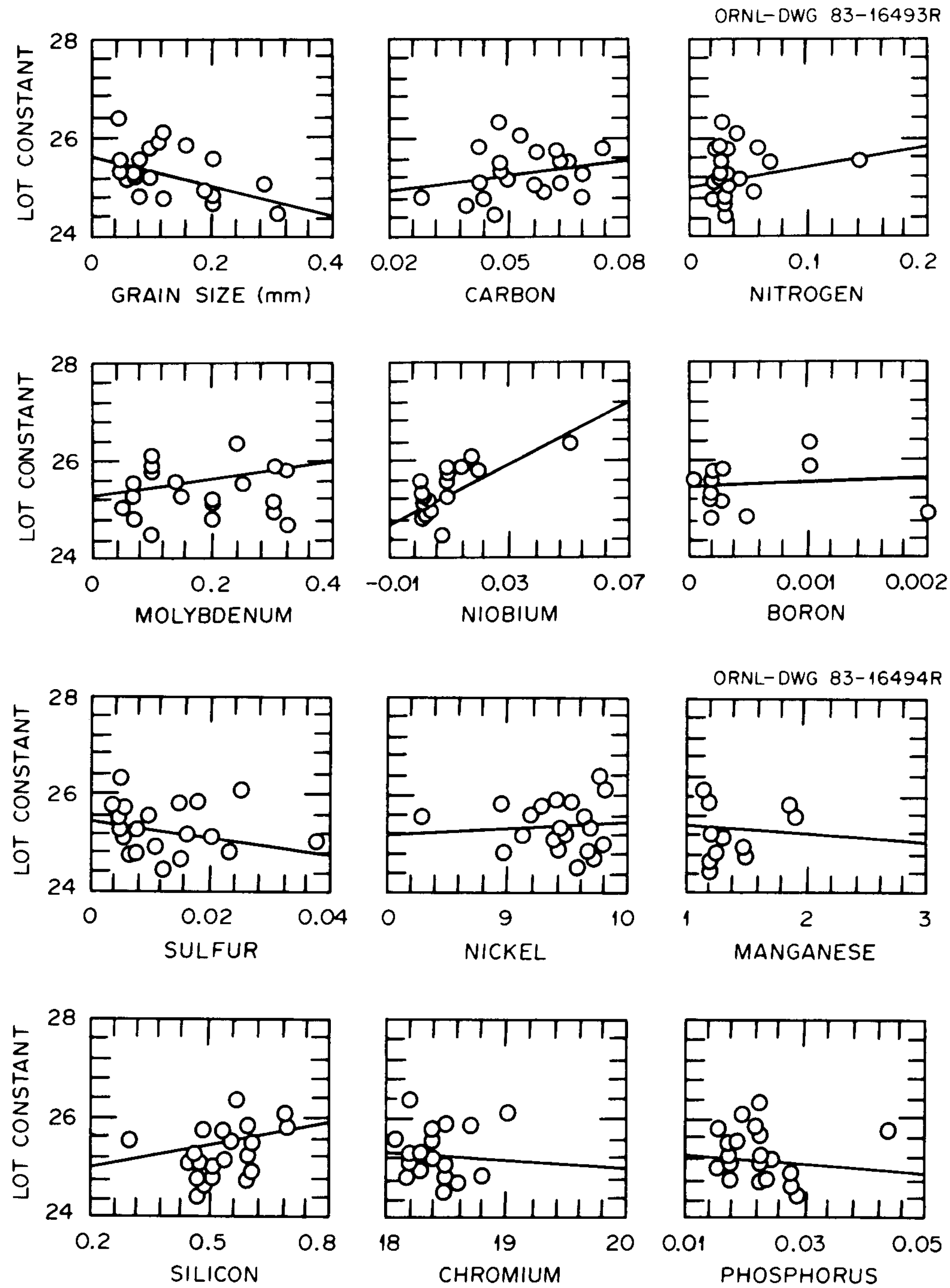

Fig. A.1. Apparent influence of various element contents (wt %) and grain size on the creep-rupture lot constant of type 304 stainless steel.

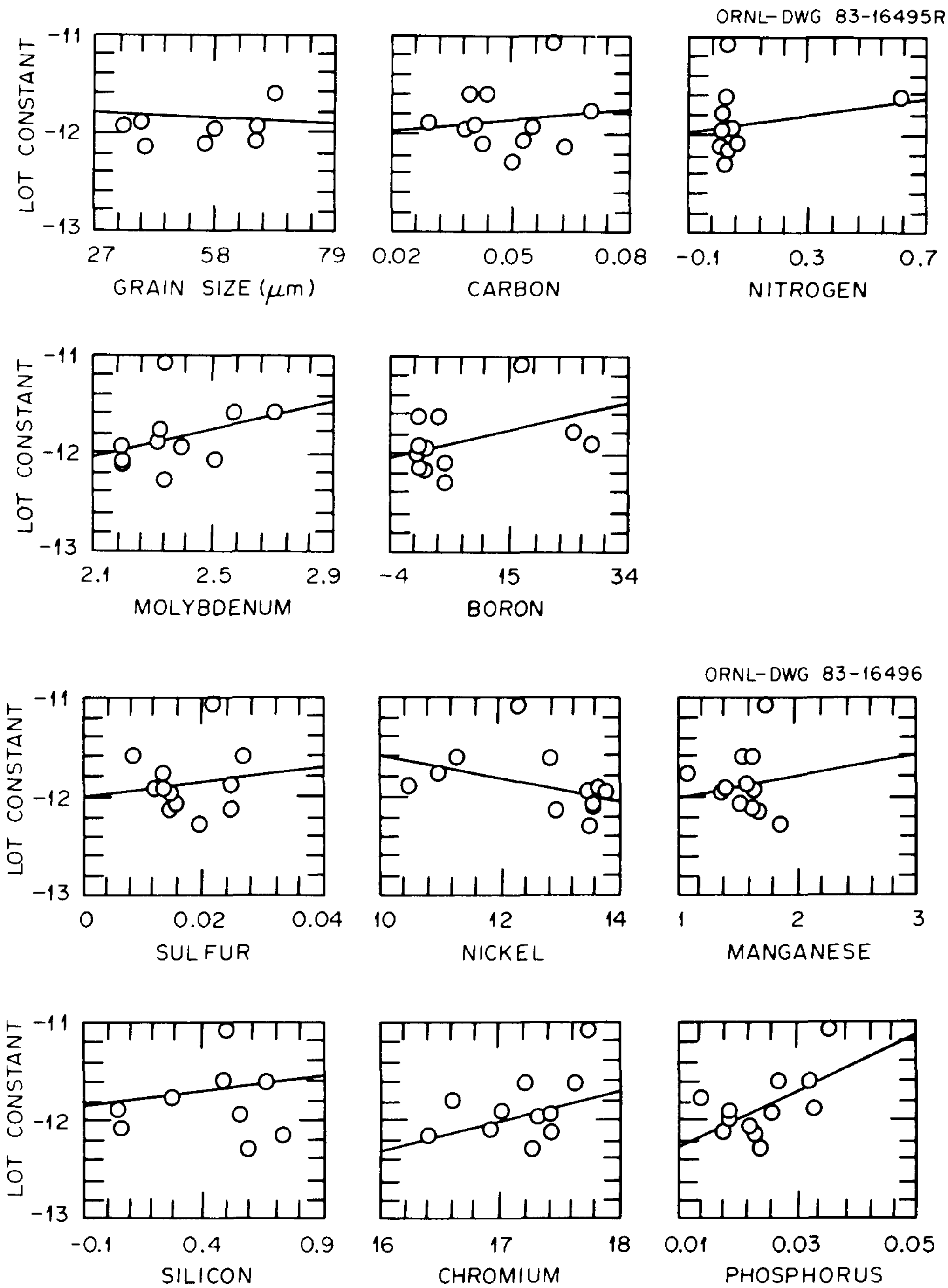

Fig. A.2. Apparent influence of various element contents and grain size on the creep-rupture lot constants of type 316 stainless steel. Contents are percentages by weight except for boron, which is in parts per million.

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