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**Evaluation of Creep and Relaxation Data
for Hastelloy Alloy X Sheet**

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EVALUATION OF CREEP AND RELAXATION DATA
FOR HASTELLOY ALLOY X SHEET

M. K. Booker

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EVALUATION OF CREEP AND RELAXATION DATA
FOR HASTELLOY ALLOY X SHEET

H. K. Booker

ABSTRACT

Hastelloy alloy X has been a successful high-temperature structural material for more than two decades. Recently, Hastelloy alloy X sheet has been selected as a prime structural material for the proposed Brayton Isotope Power System (BIPS). The material also sees extensive application in the High-Temperature Gas-Cooled Reactor (HTGR). Design of these systems requires a detailed consideration of the high-temperature creep properties of this material. Therefore, available creep, creep-rupture, and relaxation data for Hastelloy alloy X have been collected and analyzed to yield mathematical representations of the behavior for design use.

Data for time to rupture were analyzed by a generalized regression model selection procedure, after which the time and strain to the onset of tertiary creep were described as functions of the rupture life itself. The total strain to creep rupture was described as a function of the rupture life and the average strain rate to rupture. Finally, the times to accumulation of various creep strains were analyzed by regression techniques similar to those used for rupture life. A numerical interpolation scheme for prediction of times for intermediate strain levels then allows formulation of an entire creep strain-time equation from 0 to 5% creep strain.

The results yield predictions for average time and strain to rupture, time and strain to the onset of tertiary creep, and creep strain-time behavior as functions of stress and temperature (in the range 649-760°C). Data were examined for nine heats of material with rupture times ranging from 10 to 29,000 h, although the results can be extrapolated to longer times (at least 10^5 h) without anticipated difficulty. Some variations in behavior were observed from heat to heat, but this behavior could not be described analytically. The results merely represent behavior in the midrange of that displayed by these heats. The equation relating creep strain to time also yields good predictions of stress-relaxation behavior when compared with available data.

Some aspects of the behavior of this material are not fully understood. These include the reasons for and possible magnitude of heat-to-heat variations in behavior, and the reasons for the complex shapes of many creep curves. Still,

the current results present a good description of average behavior based on available data. When one accounts for uncertainties, these results should provide useful representations of behavior for design.

INTRODUCTION

Hastelloy alloy X (nominally 47 Ni-22 Cr-19 Fe-9 Mo wt %) has been successfully used in high-temperature applications for more than two decades. The success of the alloy can be attributed to its good elevated-temperature strength properties and to its excellent resistance to oxidation to temperatures of about 982°C (1800°F) or higher. It is essentially a single-phase face-centered cubic alloy, obtaining its strength primarily through solid-solution strengthening from the elements Cr, Mo, and W. Carbide precipitates may also contribute to the strengthening.

Recently, Hastelloy alloy X sheet has been selected as a prime structural material for the Brayton Isotope Power System (BIPS) being developed by AllResearch Corporation. In support of this development, we have undertaken an analysis of available creep and creep-rupture data for this material. Results include analytical descriptions for the time and strain to rupture, time and strain to tertiary creep, and creep strain-time behavior. A limited amount of stress-relaxation data have also been collected to provide a check of the creep predictions under that loading condition. These analyses comprehensively describe the available creep and creep-rupture data for this material for use in detailed design calculations.

DATA

All data examined in this report were obtained from the Stellite Division of the Cabot Corporation, both published in ref. 1 and as privately supplied by J. W. Tackett. Table 1 gives the chemical analyses of the heats for which data were available, while Table 2 gives room-temperature tensile, hardness, and grain size data for those heats.

Table 1. Elemental Composition of Heats of Hastelloy Alloy X

Heat	Composition, wt %																			
	Cr	W	Fe	C	Si	Co	Ni	Mn	Mo	P	S	Pb	Cu	Zr	Al	B	Mg	Ti	V	
4210	21.2	0.62	18.4	0.07	0.20	1.88	47.6	0.45	9.16	0.018	0.005	0.0011	0.011	0.10	<0.01	0.34	0.003	0.040	<0.01	0.04
4238	22.05	0.36	18.6	0.08	0.51	1.45	47.6	0.62	9.0	0.015	0.005	0.0011	0.012	0.07	0.01	0.27	0.002	0.045	0.01	0.05
4410	21.9	0.52	19.3	0.08	0.48	1.80	45.9	0.63	8.9	0.014	0.005	0.0012	0.011	0.15	<0.01	0.16	0.002	0.042	<0.01	0.05
4248	21.4	0.36	19.24	0.08	0.42	1.53	45.53	0.59	8.98	0.012	0.005	0.0010	0.011	0.09	<0.01	0.26	0.001	0.035	<0.01	0.04
4486	21.8	0.47	18.9	0.08	0.55	2.25	47.8	0.64	8.85	0.011	0.005	0.001	0.011	0.07	0.01	0.12	0.002	0.035	<0.01	0.08
4435	21.4	0.50	19.1	0.06	0.38	2.20	46.8	0.62	9.15	0.018	0.005	0.0013	0.012	0.11	0.01	0.26	0.002	0.043	<0.01	0.06
4392	21.7	0.42	18.7	0.07	0.49	1.95	47.5	0.59	8.85	0.005	0.005	0.0011	0.012	0.10	<0.01	0.26	0.002	0.039	<0.01	0.08
4447	21.4	0.53	19.0	0.08	0.48	2.00	47.5	0.58	9.0	0.012	0.004	0.0012	0.012	0.13	0.01	0.03	0.002	0.041	<0.01	0.05
4506	21.45	0.50	18.4	0.08	0.50	1.80	48.2	0.62	8.9	0.014	0.004	0.0014	0.012	0.08	0.01	0.21	0.002	0.047	<0.01	0.09

Table 2. Production Control Test Data

Heat	Thickness, mm (in.)	Room Temperature Tensile Properties			Rockwell Hardness	ASTM Grain Size
		Strength, MPa (Psi)		Elongation (%)		
		Ultimate Tensile	Yield			
4210	1.60 (0.063)	711.5 (103.2)	339.2 (49.2)	50.0	B87	3.5-4.5
4438	0.58-0.69 (0.023-0.027)	750.8 (108.9)	329.6 (47.8)	44.5	B82	5.5
4410	1.14-1.27 (0.045-0.050)	792.9 (115.0)	392.3 (56.9)	41.5	B87	6-6.5
4248	0.56-0.69 (0.022-0.027)	730.2 (105.9)	329.6 (47.8)	44.5	B77	5.5-6
4406	0.97-1.09 (0.038-0.043)	768.1 (111.4)	376.5 (54.6)	44.5	B85	5-5.5
4455	1.30 (0.048)	753.6 (109.3)	354.4 (51.4)	48.5	B88	5.5-6
4392	1.14-1.27 (0.045-0.050)	790.1 (114.6)	388.9 (56.4)	43.5	B89	6.5
4447	0.64 (0.025)	803.2 (116.5)	367.5 (53.3)	41.5	B80	5-6
4306	1.60 (0.063)	766.7 (111.2)	375.8 (54.5)	41.5	B91	5-6

Data used are tabulated in Appendix A. All lots of material were tested in the mill-annealed condition - solution annealed at 1177°C (2150°F) then cooled rapidly. This experimental procedure is given in detail elsewhere.¹

RESULTS

Rupture Life

As a first step in analyzing the creep behavior of Hastelloy alloy X, we sought to express the rupture life (t_r) as a function of stress (σ) and temperature (T). Many stress-rupture data are available for a variety of materials, and several methods have been proposed for the analytical treatment of those data. We chose to use a generalized regression approach^{2,3} similar to that recently applied⁴⁻¹¹ to the analysis of creep and creep-rupture data for several other materials.

A preliminary graphical analysis of the available data for individual heats showed several important trends. First, heat-to-heat variations in strength do exist, but all five primary heats of sheet material did appear to display the same general behavioral trends. Data were available from 649 to 871°C (1200–1600°F), as illustrated in Fig. 1. The data above 760°C are outside the relevant design range for BIPS or for most HTGR applications, so their use was questionable. Subsequent preliminary regression fits did indeed show that these data caused unnecessary biases in the fits of the more relevant low-temperature fits, so they were dropped from the final analyses. This trend continued for all properties examined, and all analyses described here treated data in the range 649–871°C (1200–1600°F) only.

Fig. 1. Effects of Temperature on the Creep-Rupture Behavior of Hastelloy Alloy X Sheet.

The remaining 47 stress-rupture data showed regularly spaced $\log \sigma$ - $\log t_r$ isotherms with little or no curvature. This fairly simple behavior pattern can be described with models including the terms $1/T$, $\log \sigma$, σ , $\log \sigma/T$, and σ/T . All possible models involving 2, 3, or 4 of these terms (plus a constant intercept term) were fit to the data and compared. This analysis thus involved 25 possible models. The fits were performed on the data base as a whole and on each of the five major Cabot heats separately. This preliminary run clearly identified five candidate models for further examination, as listed in Table 3. Fits of these models to available data are summarized in Table 4 in terms of the coefficients of determination, R^2 . Given in percent, R^2 is the percentage of variations in the data that can be described by a given model.

In terms of fits to the multiple-heat data, all of the models are essentially identical. The models are also nearly identical in terms of fits to the single-heat data, although model 2 has a slight overall advantage. Examination of residual plots and graphical comparisons of predictions from experimental data (and of behavior extrapolated to the estimated BIPS design life of seven years) again showed no clear advantages of any of the models over any of the others. The models, in fact, generally predict very similar results. For this reason any of the models could be chosen as optimum. Model 2 does fit slightly but not significantly better for the single-heat data. Moreover, being linear in $\log \sigma$ - $\log t_r$, it is

Table 3. Candidate Models for Analysis of Time to Rupture for Hastelloy Alloy X

Model	Form
1	$\log t_r = a_0 + a_1/T + a_2 \log \sigma + a_3 \sigma$
2	$\log t_r = a_0 + a_1/T + a_2 \log \sigma + a_3 \log \sigma/T$
3	$\log t_r = a_0 + a_1/T + a_2 \log \sigma + a_3 \sigma/T$
4	$\log t_r = a_0 + a_1/T + a_2 \sigma + a_3 \log \sigma/T$
5	$\log t_r = a_0 + a_1/T + a_2 \log \sigma/T + a_3 \sigma/T$

Table 4. Fits of Candidate Models to Data for Time to Rupture for Hastelloy Alloy X

Heat	Number of Data	Coefficient of Determination, R^2 (%), for Model				
		1	2	3	4	5
4210	11	98.4	98.8	98.4	98.8	98.8
4248	8	99.2	99.6	99.3	99.6	99.6
4410	8	93.8	95.6	94.0	94.3	94.5
4438	12	93.4	92.7	93.1	92.3	92.3
4486	8	97.0	93.0	97.1	97.8	97.8
Combined	47	88.6	89.5	88.6	89.2	89.2

easier to work with and is less susceptible to biases introduced by fitting multiple-heat data (see Appendix B). Therefore, we chose this model to establish the predicted seven-year creep-rupture strength of Hastelloy alloy X.

The predictions from model 2, as fitted to each of the five individual heat data sets, are compared with the corresponding data in Figs. 2-6. For comparison, average isotherms were also graphically drawn through those data. Table 5 compares the analytical and graphical results. Figure 7 compares the predictions from model 2 as fitted to the combined multiple-heat data set with those data. Also shown are estimated "lower limit" predictions. Appendix C briefly describes some methods for such limit setting. We concluded from these data that such limits could be adequately set by subtracting twice the standard error of estimate (SEE) from the fit of model 2 from the expected value of $\log t_r$ from that model. Here, SEE is 0.26. Thus, the expected behavior is given by

$$\log t_r = -24.51 + 40,030/T + 4.248 \log \sigma - \frac{10,100}{T} \log \sigma, \quad (1)$$

and the "lower limit" behavior by

$$\log t_r' = -25.03 + 40,030/T + 4.248 \log \sigma - \frac{10,100}{T} \log \sigma, \quad (2)$$

where

T = temperature (K), and

σ = stress (MPa).

Note that this method of limit setting has no precise statistical meaning. It is merely a convenient means to express the width of the scatter band for the available data.

Fig. 2. Comparison of Experimental and Predicted (from Individual Heat Fits) Rupture Life for Hastelloy Alloy X Heat 4210.

Fig. 3. Comparison of Experimental and Predicted (from Individual Heat Fits) Rupture Life for Hastelloy Alloy X Heat 4248.

Fig. 4. Comparison of Experimental and Predicted (from Individual Heat Fits) Rupture Life for Hastelloy Alloy X Heat 4410.

Fig. 5. Comparison of Experimental and Predicted (from Individual Heat Fits) Rupture Life for Hastelloy Alloy X Heat 4438.

Fig. 6. Comparison of Experimental and Predicted (from Individual Heat Fits) Rupture Life for Hastelloy Alloy X Heat 4486.

Table 5. Estimated Values of Seven-Year Creep-Rupture Strength of Hastelloy Alloy X Sheet^a

Data Set	Temperature, °C (°F)		
	649 (1200)	760 (1400)	871 (1600)
Analytical Results (Model 2)			
4210	147	62	25
4248	118	46	16
4410	137	55	13
4438	102	40	18
4486	116	46	16
Combined	128	51	18
Graphical Results			
4210	145	65	24
4248	118	45	16
4410	124	63	13
4438	111	48	18
4486	117	58	14
Combined	122	54	17

^aTabulated values represent estimated seven-year creep-rupture strengths in MPa.

Fig. 7. Comparison of Experimental Multiple-Heat Data for Rupture Life with Average Predictions and Average -2SEE.

The multiple-heat predictions are meant to describe approximately average behavior as displayed by the various heats for which data are available. That the predicted strengths are indeed about average for these heats can be checked conveniently by a "strength trend curve"¹² constructed for the data. As illustrated in Fig. 8, such a curve involves plots of the estimated strength values for individual heats as a function of temperature. For the current individual heat data sets, sufficient information was available for reasonable estimation of seven-year creep-rupture strengths for each heat. However, lack of data does in some cases introduce uncertainty, as witnessed by differences between the analytical and graphical predictions in Table 5. Therefore, we probably cannot rely on such a strength trend curve for prediction in this case. The curves can be conveniently compared with the above multiple-heat fits. The fits appear adequately consistent with the trend curves.

Fig. 8. Comparison of Seven-Year Creep-Rupture Strengths Estimated by Single-Heat Fits with Predictions from Multiple-Heat Fits.

Another way to treat multiple-heat data is to treat the heat-to-heat variations directly by including some index of heat-to-heat variability in the models.^{2,3} Attempts to identify such an index for the current data proved unsuccessful, as witnessed by Fig. 9-13. In Fig. 9 the seven-year creep-rupture strength does not significantly correlate with room-temperature yield strength. Creep strength for several materials^{6-9,13-15} has been found to be correlated with elevated-temperature ultimate tensile strength (UTS). No such data were available for the current material, but with the exception of heat 4210 (the thickest sheet with the largest grains), room-temperature ultimate tensile strength correlated positively with creep-rupture strength at 649 and 760°C (Fig. 10), but at 871°C, rupture

Fig. 9. Relationship Between Seven-Year Creep-Rupture Strength and Room-Temperature Yield Strength for Hastelloy Alloy X Sheet.

Fig. 10 Relationship Between Seven-Year Creep-Rupture Strength and Room-Temperature Ultimate Tensile Strength for Hastelloy Alloy X Sheet.

Fig. 11. Relationship Between Seven-Year Creep-Rupture Strength and Sheet Thickness for Hastelloy Alloy X Sheet.

Fig. 12. Relationship Between Seven-Year Creep-Rupture Strength and Chemical Composition for Hastelloy Alloy X Sheet.

Fig. 13. Relationship Between Seven-Year Creep-Rupture Strength and Grain Size for Hastelloy Alloy X Sheet.

strength appeared to decrease with increasing room-temperature UTS. Rupture strength appeared to relate to sheet thickness at 649 and 760°C, but not at 871°C, as shown in Fig. 11. Figure 12 illustrates attempts to relate rupture strength to chemical composition. The concentrations (wt %) of the primary solution-strengthening elements (weighted to be approximately the same magnitude) appear to relate to rupture strength at 649 and 760°C, but not at 871°C. Finally, Fig. 13 shows that rupture strength can be related to grain size at 871°C, but not at the lower temperatures. The net conclusion from analyses such as illustrated in Figs. 9-13 was that no adequate simple index of heat-to-heat variability could be identified with available information. A more sophisticated statistical approach might be that outlined by Sjodahl,¹⁶ but confidence in the results increases too little to justify the increased complexity of the analysis. Therefore, the results given by Eqs. (1) and (2) seem to best represent the creep-rupture behavior of Hastelloy alloy X sheet at this time.

For subsequent discussions of the behavior of this alloy, realize that other heats from other vendors or of different product forms may display behavior that ranges beyond that shown in these Cabot heats of sheet material.

Strain to Rupture

High-strength materials such as Hastelloy alloy X sometimes display poor long-term creep ductility. To assess this possibility, we examined the total elongation to rupture values from the above creep-rupture data for this material. As creep ductility data often require,¹⁷ we had to examine the data on individual heats to identify trends. A common way of plotting such data is illustrated in Fig. 14, where the strain* to rupture, ϵ_r , is plotted against the logarithm of the rupture life. However, trends from such plots appeared confusing. Plotting ϵ_r against the creep stress gave more meaningful results consistent with those found previously¹⁷ for

*All strain values in this report represent engineering strains rather than true strains.

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Fig. 14. Strain to Rupture as a Function of Rupture Life For Hastelloy Alloy X Heat 4438.

several other materials. Examination of these plots (Figs. 15-19) lead to the following qualitative conclusions (on which the hand-drawn dashed curves in the figures are based): (1) At a given stress, higher temperatures (in the range 649-871°C) tend to correspond to higher values of rupture strain. (2) In general, the rupture strain tends to decrease as stress decreases at a given temperature. However, the data indicate a peaking effect at 649°C, where the rupture strain begins to decrease again as stress increases past a certain critical value. The data at 760 and 871°C indicate that a similar effect may also occur there at stresses higher than those for which data are available. (3) All values of e_r in the current data range fall above 25%, but if trends in the data continue to lower stresses, somewhat lower values of strain to rupture might be experienced at stresses at or below the estimated seven-year creep-rupture strengths given in Table 5. Table 6 displays these values as determined from Figs. 15-19.

Fig. 15. Strain to Rupture as a Function of Creep Stress for Hastelloy Alloy X Heat 4210.

Fig. 16. Strain to Rupture as a Function of Creep Stress for Hastelloy Alloy X Heat 4248.

Fig. 18. Strain to Rupture as a Function of Creep Stress for Hastelloy Alloy X Heat 4438.

Fig. 17. Strain to Rupture as a Function of Creep Stress for Hastelloy Alloy X Heat 4410.

Fig. 19. Strain to Rupture as a Function of Creep Stress for Hastelloy Alloy X Heat 4486.

Table 6. Estimated Values of Total Elongation to Rupture Corresponding to a Rupture Life of Seven Years

Heat	Elongation, (%) ^b		
	649 (1200)	760 (1400)	871 (1600)
4210	40.5	27.0	16.5
4248	21.5	8.0	7.5
4410	45.5	28.0	9.5
4438	22.5	12.5	12.0
4486	22.0	15.0	7.0

^aElongation values estimated graphically.

^bFor temperatures in °C (°F).

Quantitative techniques for prediction of creep ductility values have recently been applied to a number of materials.^{4-6,17-18} The analysis is based on a concept introduced by Smith¹⁹ defining the average strain rate to rupture, $\dot{\epsilon}_r$, by

$$\dot{\epsilon}_r = e_r / t_r . \quad (3)$$

The quantities $\dot{\epsilon}_r$ and t_r both typically display less scatter and variability than e_r itself. Thus, those quantities can be analyzed separately (as proposed by Goldhoff²⁰), and e_r is given as their product. For the current data, $\dot{\epsilon}_r$ was analyzed as above for t_r . The combined multiple-heat data were again described by model 2 as

$$\log \dot{\epsilon}_r = 24.02 - 39,080/T - 2.666 \log \sigma + \frac{9,140}{T} \log \sigma . \quad (4)$$

Figure 20 compares the predictions for e_r from Eqs. (1) and (4) with experimental data. The analysis captures many of the essential features

Fig. 20. Comparison of Experimental and Predicted Values for Strain to Rupture.

of the stress and temperature dependences of e_p , but some of the details (such as the peaking at 649°C) are lost because combining the heats increases the scatter. Figures 15-19 and Table 6 thus probably give the best information on expected trends in the creep-rupture ductility of Hastelloy alloy X sheet.

Onset of Tertiary Creep

Analysis of the onset of tertiary creep is complicated by the problem of defining the onset point for individual curves relating creep strain and time. For a classical curve, as shown in Fig. 21, the times t_2 and t_{ss} both commonly measure the time to the onset of tertiary creep, while e_2 and e_{ss} are the corresponding creep strains. The offset values t_{ss} and e_{ss} will be used in this report. The definition problem becomes much more complicated as the creep curve shape becomes less classical. In such cases an increasing creep rate may not signal tertiary creep in the sense of impending instability or failure. Rather, such an increase may merely indicate a softening or decreased resistance to creep flow in the material.⁸⁻¹⁰

Fig. 21. Schematic Definition of the Onset of Tertiary Creep in a Classical Creep Curve.

The determination of the tertiary creep behavior of Hastelloy alloy X sheet is also complicated by the form of the available information. The curves supplied by Cabot-Stellite for heats 4210 and 4486 were plotted to, or almost to rupture, resulting in coarse strain scales that perhaps camouflaged the above nonclassical creep behavior. Strain scales for the other heats were generally expanded enough for detailed analysis of the initial portions of the curves, but the curves often covered too little time for the end of the second linear creep portion to be determined.

Many of the current creep curves for Hastelloy alloy X sheet appeared nonclassical. A typical nonclassical curve is shown in Fig. 22. As seen in the figure, determination of the onset of tertiary creep can involve considerable judgment. These curves exhibit an initial primary creep region that tends toward a linear secondary creep region. Then the creep rate increases and a second region of linear creep is observed. A second "primary" creep region — a region of decreasing creep rate — may or may not occur after the creep rate increase and before the second linear creep

Fig. 22. Typical Nonclassical Creep Curve Observed in Hastelloy Alloy X Sheet.

region. This creep rate increase occurs early in life (often at $t < 0.1 t_r$) and at creep strains of about 1% or less. These nonclassical curves seemed most common for heats 4410, 4438, and 4210, the highest incidence occurring for high stresses at 649°C and for low stresses at 871°C. However, many curves were borderline — appearing to represent a transition between classical curves and these nonclassical curves.

This behavior is phenomenologically quite similar to that observed^{8,9,21} in 2 1/4 Cr-1 Mo steel, in which the increasing creep rates before a second linear stage were related to a change in creep mechanism due to metallurgical changes in the material. In the current case there is insufficient information to develop a physical explanation for the observed nonclassical creep behavior. However, it appears probable that the increasing creep rates immediately following the first linear stage do not represent onset of tertiary creep. Tertiary creep data is widely analyzed^{18,22-25} by examining the relationship between the time to tertiary creep and the corresponding time to rupture. Figure 23 illustrates this relationship both for the classical curves and the nonclassical curves. The end of the second linear

Fig. 23. Relationship Between Time to Onset of Tertiary Creep and Rupture Life.

stage in the nonclassical curves appears to correlate much better with the classical onset of tertiary creep than does the end of the first linear stage. The creep strain at the onset of tertiary creep for the available classical curves was always 2% or greater, as was the strain to the end of the second linear stage in the nonclassical curves. However, the strain to the end of the first linear nonclassical stage was typically 1% or less. From these observations the first region of increasing creep rates in the nonclassical curves apparently does not represent the onset of tertiary creep. Rather, this onset is signaled by the increasing creep rates at the end of the second linear creep stage.

From Fig. 23 the time to the onset of tertiary creep, t_{ss} , is given by $0.4 t_r$. Trends in $\dot{\epsilon}_{ss}$ appear confusing even when viewed on a single-heat basis. Following the above analysis for $\dot{\epsilon}_r$, we define the average creep rate to tertiary creep, $\dot{\epsilon}_{ss}$, by $\dot{\epsilon}_{ss}/t_{ss}$. As shown in Fig. 24, $\dot{\epsilon}_{ss}$ was related to t_r by

Fig. 24. Relationship Between Average Creep Rate to Tertiary Creep and Rupture Life.

$$\dot{e}_{ss} = 60 t_r^{-1.2}, \quad (5)$$

so that

$$e_{ss} = 24 t_r^{-0.2}. \quad (6)$$

Equation (6) describes the available data for e_{ss} reasonably well within the limits of data scatter, even though trends in those data are by no means clear. Note that expressing e_{ss} simply as a function of t_r implies that for a given temperature, e_{ss} decreases as stress decreases and that for a given stress, e_{ss} decreases as temperature decreases. These trends resemble those noted above for e_r . Equation (6) predicts an average e_{ss} of 2.65% for a seven-year rupture life. This value would in actuality vary from heat to heat, but this value should represent a compromise based on the five major heats for which data are currently available.

Creep Strain-Time Behavior

Various methods for analytically describing the accumulation of creep strain with time are discussed in ref. 6. Two methods have been particularly popular in recent years. The first involves the use of "creep equations," direct analytical expressions for creep strain as a function of time, stress, and temperature. To begin this analysis,^{4-11,26} usually some basic analytical relation between strain and time is postulated, such as the rational polynomial creep equation.²⁷ The parameters in this equation are then themselves expressed as functions of stress and temperature. This approach allows considerable flexibility and is generally viable if the creep curves under various conditions do indeed display shapes that can all be described by the same strain-time equation form.

Alternatively, the times (t_x) to accumulation of given creep strains can be analyzed by techniques similar to those commonly used for rupture life data. This method was used to describe data for Incoloy alloy 800 (grade 2), Inconel alloy 625 (grade 2), and 2 1/4 Cr-1 Mo steel by Roberts and Sterling.²⁸ Using a series of strain levels and interpolating between them, one can then reconstruct an entire creep curve. This method suffers from the fact that it does not directly yield an expression for the creep

strain rate, which is required in design calculations. It can be used to generate isochronous stress-strain curves such as those given in ASME Code Case 1592.²⁹ This method was extended by Sterling³⁰ for Alloy 800 and 2 1/4 Cr-1 Mo steel by assuming a power law equation relating strain level to time and thus analytically "tying together" the various strain levels. The Sterling equation can thus yield an expression for creep rate as well as for creep strain, and is in fact a form of creep equation.

The Sterling approach suffers from lack of flexibility because it assumes that an entire creep curve can be described by a single strain-time power law ($\epsilon = At^n$), that the times to given creep strains can all be described by a Larson-Miller³¹ time-temperature parameter, and that the power law exponent, n , is independent of stress. Recent work by Ellis³² alleviates many of the problems, however. Using a Larson-Miller parameter for the time to various strain levels, Ellis merely assumed a power law strain-time equation between strain levels. This piecewise approach adds considerable flexibility, and the method does not require use of Larson-Miller or any other particular form for $t_x(\sigma, T)$. Further, Ellis proposed an interpolation scheme that breaks the curve into smaller and smaller increments to give creep rate predictions sufficient continuity for the method to be used in finite element design codes.

Application of the classical creep equation method to the current data for Hastelloy alloy X sheet immediately presents two difficulties. First, many of the available curves are plotted such that a detailed analysis of the initial primary stage, quite small compared with the entire creep curve, is not possible. Second, the nonclassical creep behavior noted above makes description of all curves with a single strain-time equation form nearly impossible. The Ellis approach does look attractive, however, especially since times tabulated to 0.1, 0.2, 0.5, 1, 2, and 5% creep strain are available in ref. 1. Analysis of these data should yield curves that generally represent compromise behavior among the curves displayed by the various heats at various stresses and temperatures.

To begin the analysis the time values, t_x , corresponding to various strain levels, were all analyzed as functions of stress and temperature

by the same regression techniques used above for rupture life. The method does not require that all t_x values be described by the same model form, but such a circumstance helps to minimize inconsistencies among the various values. For creep strains of 1, 2, and 5%, model 2 was again the best model. For the smaller strains model 1 was generally better, but model 2 was only somewhat worse (in terms of R_2 values). Within the limitations of a seven-year total life, the two models do not yield substantially different behavior, and model 2 was chosen to represent all of the t_x values. Thus,

$$\log t_{0.1\%} = -10.93 + 19,760/T - 2.09 \log \sigma - \frac{1,874}{T} \log \sigma , \quad (7)$$

$$\log t_{0.2\%} = -14.59 + 25,360/T - 0.524 \log \sigma - \frac{4,200}{T} \log \sigma , \quad (8)$$

$$\log t_{0.5\%} = -12.94 + 24,960/T - 1.28 \log \sigma - \frac{3,824}{T} \log \sigma , \quad (9)$$

$$\log t_{1\%} = -15.89 + 29,750/T - 0.398 \log \sigma - \frac{5,377}{T} \log \sigma , \quad (10)$$

$$\log t_{2\%} = -21.11 + 36,350/T - 1.96 \log \sigma - \frac{8,200}{T} \log \sigma , \quad (11)$$

and

$$\log t_{5\%} = -22.68 + 38,520/T - 2.63 \log \sigma - \frac{8,936}{T} \log \sigma . \quad (12)$$

Table 7 summarized the fits of these equations to available data, while Figs. 25-30 compare the predictions of the equations with experimental data. A compromise SEE value of 0.28 log h can be selected to calculate lower and upper limit values by $t_x \pm 2$ SEE. Note that Appendix D discusses some recent alternative analyses for $t_{1\%}$.

Table 7. Summary of Fits to Data for Time to Various Levels of Creep Strain

Creep Strain (%)	Number of Data	R^2 , (%) ^a	SEE ^b
0.1	70	86.2	0.28
0.2	73	86.8	0.30
0.5	73	89.9	0.28
1.0	73	93.1	0.26
2.0	62	91.6	0.28
5.0	52	90.2	0.28

^a Coefficient of determination. R^2 in percent gives the percentage of variations in the data described by a given model.

^b Standard error of estimate in terms of $\log t_x$.

Fig. 25. Comparison of Experimental and Predicted Values of Time to the Accumulation of 0.1% Creep Strain.

Fig. 26. Comparison of Experimental and Predicted Values of Time to the Accumulation of 0.2% Creep Strain.

Fig. 27. Comparison of Experimental and Predicted Values of Time to the Accumulation of 0.5% Creep Strain.

Fig. 28. Comparison of Experimental and Predicted Values of Time to the Accumulation of 1% Creep Strain.

Fig. 29. Comparison of Experimental and Predicted Values of Time to the Accumulation of 2% Creep Strain.

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Fig. 30. Comparison of Experimental and Predicted Values of Time to the Accumulation of 5% Creep Strain.

Equations (7-12) can be converted to a complete strain-time curve prediction by assuming that a power law equation can be used to interpolate between strain levels. The curve between 0.1 and 0.2% strain can be extrapolated to strain values less than 1%. Extrapolation to strain values higher than 5% is not recommended. This scheme is equivalent to linear interpolation between strain levels in $\log t$ - $\log \epsilon$ space. Thus, the power law parameters can be evaluated very simply. For instance, between $\epsilon = 0.1\%$ and $\epsilon = 0.2\%$ the parameters are

$$n = \frac{\log 0.2 - \log 0.1}{\log t_{0.2\%} - \log t_{0.1\%}}, \quad (13)$$

and

$$A = \frac{0.1}{t_{0.1\%}^n}. \quad (14)$$

The parameters in the other strain ranges can be evaluated analogously.

Figures 31-36 compare predicted average and upper and lower limit creep behavior with experimental data (on log-log coordinates for convenience). In general, the predictions describe the behavior well. One inconsistency in the predictions manifests itself in Fig. 32 at low strains. All the equations for various strain levels yield straight line $\log \sigma$ - $\log t_x$ isotherms that converge at high stresses. At 649°C the convergence between the 0.1 and 0.2% isotherms causes anomalous strain-time predictions above stresses of about 200 MPa. This anomaly might be removed analytically, but stresses above 200 MPa at 649°C are probably outside the range that would be encountered in BIPS service, so this limitation should not be important. If predictions were needed in that range, they could be made by ignoring the 0.2% predictions and using $t_{0.1\%}$ and $t_{0.5\%}$ to construct the initial power law segment. The peak applicable stress at 760°C is about 160 MPa, while that at 871°C is about 120 MPa. Peak stresses at intermediate temperatures can be established by linear interpolation.

Fig. 31. Comparison of Experimental and Predicted Creep Strain-Time Behavior at 207 MPa and 649°C.

Fig. 32. Comparison of Experimental and Predicted Creep Strain-Time Behavior at 276 MPa and 649°C.

Fig. 33. Comparison of Experimental and Predicted Creep Strain-Time Behavior at 138 MPa and 760°C.

Fig. 34. Comparison of Experimental and Predicted Creep Strain-Time Behavior at 69 MPa and 760°C.

Fig. 35. Comparison of Experimental and Predicted Creep Strain-Time Behavior at 69 MPa and 871°C.

Fig. 36. Comparison of Experimental and Predicted Creep Strain-Time Behavior at 28 MPa and 871°C.

At each point where two power laws are joined, the resulting discontinuity in creep rate will present difficulties when one attempts to use the equations for most finite element design codes. This problem can be surmounted as proposed by Ellis.³² However, a much simpler and more satisfying approach is to simply interpolate among the various t_x values with a scheme for continuous creep rates. Such an interpolation can be accomplished with any number of common numerical schemes. We suggest the use of a cubic spline interpolation procedure, which will provide continuity in both $\dot{\epsilon}$ and $\ddot{\epsilon}$. The interpolation should be performed in log-log space since the strain levels used are approximately evenly spaced on log coordinates.

The above creep equation can be used in the temperature range from 649 to 871°C for strains up to 5%. Some caution should be used, however, since this strain value will occur beyond the onset of tertiary creep for rupture lives greater than about 2500 h. To ensure safe design operation, the time to tertiary creep, t_{ts} , should be an upper limit on the applicability of the creep equation for design.

Relaxation Data

Relaxation data serve three purposes. First, in the absence of any other data, charts of relaxed stress as a function of time and temperature help the designer estimate the secondary stress levels in components. Second, when creep laws and constitutive equations are available, the designer may use relaxation data to verify his method of predicting creep response under variable stresses. Third, the data on the history of stress and strain rate — data derived from relaxation tests — can be used to develop constitutive equations by methods outlined by Hart.³³ Several relaxation tests on Hastelloy alloy X yielded data to assess the ability of proposed creep equations (such as that discussed in the last section) to predict creep and relaxation under varying stress at temperature.³⁴

The first series of relaxation tests were performed on 1.6-mm sheet specimens at 632°C. Four successive relaxation "runs" were performed on the same specimen. The results of the relaxation tests are shown in Fig. 37. We see that the relaxation strength increases from one run to

Fig. 37. Comparison of Experimental and Predicted Stress Relaxation Behavior of Hastelloy Alloy X Sheet at 732°C.

the next. This suggests that creep hardening occurs for these test conditions. Also shown in Fig. 37 are predicted relaxation curves from the creep equation developed in the previous section of this report. The hypothesis of strain hardening³⁵ was used in conjunction with the creep equation to make relaxation predictions. Creep strain hardening was assumed to accumulate from one run to the next. The predictions agree well with the data, especially considering that the material tested in relaxation differed somewhat from that tested in creep. The predictions also show hardening from one run to the next. These results further assure the applicability of the current creep equation for design use. They do not, however, imply that the strain hardening scheme applied here is applicable in general. At this time the data are insufficient to experimentally verify any hardening laws.

Relaxation tests at temperatures from 593 to 816°C (1100 to 1500°F) have been conducted on Hastelloy alloy X bar material. These curves are represented in terms of stress remaining after a given relaxation time as a function of temperature in Fig. 38. The figure also shows predictions from

Fig. 38. Comparison of Experimental Relaxation Strength for Bar Material from 593 to 816°C with Strengths Predicted from Creep Equation and Strain Hardening.

the current creep equation and strain hardening. The predictions again compare well with the data. More data under different conditions are needed, including data to assess hardening rules, history effects, etc.

DESIGN CRITERIA

Specific design criteria and safety factors for a particular system depend on the nature of the system and on the potential consequences of failure. However, several material properties are of general interest for developing design criteria under creep conditions. These three criteria, all used in ASME Code Case 1592,²⁸ are creep rupture, onset of tertiary creep, and the accumulation of a given amount of creep strain.

Actual creep rupture is an obvious design criterion in that it directly relates to failure. In some cases the onset of tertiary creep may also relate to failure. Penny and Marriott³⁶ note that creep rupture as an instability relates more to the conditions preceding and causing the onset of the instability than to the conditions following the instability. Therefore, the onset of tertiary creep, if viewed as the beginning of failure, can have fundamental significance. Grant et al.^{37,38} support this view, having found that the onset of tertiary creep — especially in terms of ductility — can be more fundamental than rupture itself.

Garofalo³⁹ indicated that the onset of tertiary creep may relate to several factors, most commonly:

1. increasing stress under constant load (due to a decrease in cross-sectional area with uniform deformation),
2. increasing stress due to necking,
3. void formation and cracking,
4. recovery,
5. recrystallization, and
6. precipitation or resolution of one or more phases.

In a given situation one or more factors could produce tertiary creep. The exact mechanism and meaning of tertiary creep appear to be complex and probably vary from material to material. Our current information does not

allow us to assess the physical meaning of the onset of tertiary creep in Hastelloy alloy X. We hope to develop such information in future studies.

The final common creep design criterion involves a so-called strain limit. In Code Case 1592,²⁸ design stresses are limited to those that will yield no more than 1% strain in the design life of a component. Certainly, strain as a design criterion has several advantages. For instance, the strain history of a structural component is often easier to predict than its stress history. Other advantages to the use of strain limit as a design criterion include the fact that such a limit (1) can be related to actual measurements on components, (2) prevents large deformations and thus ensures the validity of analytical techniques for low deformation, (3) limits deformation due to ratchetting, (4) might easily apply to weldments, and (5) provides a means of treating creep fatigue failure.

The choice of 1% as the limiting strain is reasonable even if somewhat arbitrary. A strain limit that relates to creep ductility might be more meaningful, but predictions of creep ductility inherently contain large assumptions. For Hastelloy alloy X the 1% limiting strain criterion appears reasonable, since it limits deformation to the region before the predicted onset of tertiary creep.

Table 8 compares the average and lower limit stresses derived from the three above criteria for a design life of seven years. The lower limit values

Table 8. Design Creep Strength of Hastelloy Alloy X
Sheet as Established by Three Criteria

Temperature		Average and (Minimum) Stress, MPa, for Seven Years to:		
(°C)	(°F)	1% Creep Strain	Rupture	Onset of Tertiary Creep
649	1200	73 (59)	127 (107)	111 (93)
704	1300	45 (36)	83 (68)	71 (58)
760	1400	28 (22)	51 (41)	43 (35)
816	1500	18 (14)	30 (24)	25 (20)
871	1600	11 (8.6)	18 (14)	14 (11)

were calculated as we described previously. They correspond to approximate minimum expected material behavior and do not in themselves address design safety. Application of appropriate factors to the values in the table would yield estimated allowable operating stresses by the various criteria.

SUMMARY

Available creep and creep-rupture data for Hastelloy alloy X sheet have been examined to develop analytical expressions for average behavior as a function of stress and temperature in the range from 649 to 871°C (1200 to 1600°F). The average time to rupture, t_r , can be described as a simple function of stress and temperature by Eq. (1) in the text. This equation predicts behavior that is approximately average among five major heats of sheet material for which data were available. Heat-to-heat variations in behavior do exist, but these variations were not found to relate simply to any convenient index such as tensile properties or chemistry. An empirical safety factor on $\log t_r$ of two standard errors of estimate yields a good approximation for lower limit behavior among the heats studied.

The strain to rupture, ϵ_r , was found to generally increase with temperature at a given stress and to increase with stress at a given temperature, although a peak in ϵ_r with stress was noted at 649°C. Actual magnitudes vary from heat to heat, but clearly, ϵ_r can easily be under 10% for a seven-year life.

Hastelloy alloy X often displays nonclassical creep curves, and determination of the onset of tertiary creep is not necessarily straightforward. Using the approach described in the text, the time to the onset of tertiary creep, t_{ss} , was described as $0.4 t_r$. The corresponding creep strain, ϵ_{ss} , can be estimated by $24 t_r^{-0.2}$, although trends again vary from heat to heat and uncertainty is large.

Creep strain behavior over time was described first by expressing the average times to 0.1, 0.2, 0.5, 1, 2, and 5% creep strain as functions of stress and temperature. Then, curves relating creep strain to time

can be predicted by interpolating between strain levels, as by use of cubic spline interpolation. The equation can be extrapolated to strains lower than 0.1% assuming a power law relationship between strain and time. Extrapolation beyond 5% creep strain is not recommended. These strain-time predictions should not be used above stresses of 200 MPa at 639°C, 160 MPa at 760°C, or 120 MPa at 816°C. Comparison of experimental relaxation data with behavior predicted from the creep equation and strain hardening yielded good results, although the initial portions of some of the relaxation curves involved stresses higher than the limits listed above.

The forms of the equations in this report allow analytical extrapolation to longer times and lower stresses than those for which data are available. Such extrapolations cannot be verified as accurate in the absence of a detailed proven physical theory for the subject processes. No such theory exists for the creep and creep-rupture of Hastelloy alloy X, so such extrapolation should be done with caution.

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APPENDIX A

Table A1. Creep-Rupture Data^a for Hastelloy Alloy X Sheet

Test	Heat	Laboratory and Orientation ^b	Stress		Loading Strain ^c (%)	Time, h. to Reach Creep Strain, % of					Time to Rupture ^d (h)	Total Elongation (%)	Note ^e	Minimum Creep Rate (1/h)	
			(MPa)	(ksi)		0.1	0.2	0.5	1.0	2.0					5.0
Tested at 649°C (1220°F)															
14468	94210	T T	300	55.0	12.80						10.0	60.6	65.0	F	0.3000
13264	4240	J T	365	50.0	3.22		0.1	0.5	2.0	6.0	10.0	69.4	17.1	F	0.3000
13265	4406	J T	365	50.0	3.36		0.1	0.4	2.5	6.0	17.0	60.0	65.0	F	0.3000
13165	4240	J T	276	40.0	0.79	1.0	4.0	12.0	20.0	30.0	70.0	326.1	79.0	F	0.0040
13797	4430	T T	276	40.0	0.36	2.5	5.2	12.5	21.0	37.0	80.0	300.0	79.1	F	0.0030
13800	4430	T T	276	40.0	0.96	2.0	4.0	12.5	23.0	39.0	87.0	300.0	70.7	F	0.0030
13164	4406	J T	276	40.0	0.69	2.0	4.9	8.0	22.5	34.0	82.0	342.4	75.6	F	0.0030
12725	94210	T T	276	40.0	1.40	2.0	4.9	15.0	29.0	52.0	124.0	374.8	91.1	F	0.0010
13266	4240	J L	207	30.0	0.19	11.0	23.0	45.0	81.0	140.0	400.0	1750.2	50.7	F	0.0120
14442	4430	T T	207	30.0	0.30	5.0	29.0	61.0	99.0	182.0	440.0	2100.4	52.4	F	0.0120
13147	4406	J L	207	30.0	0.21	6.8	21.0	40.0	90.0	180.0	440.0	1630.1	51.2	F	0.0110
13167	4240	J T	173	25.0	0.16	23.0	46.0	85.0	190.0	400.0	1400.0	(1000.0)		D	0.00320
13703	4430	T T	173	25.0	0.18	25.0	45.0	100.0	210.0	490.0	1300.0	770.3	5	F	0.00200
12566	4430	T T	173	25.0	0.16	37.0	65.0	128.0	220.0	450.0		(301.3)		D	0.00400
13148	4406	J T	173	25.0	0.18	37.0	67.0	145.0	295.0	640.0	1900.0	(1600.0)		D	0.0020
12716	94210	T T	173	25.0	0.09	117.0	175.0	209.0	440.0	950.0	1200.0	(4062.0)		X	0.00210
14358	4430	T L	130	20.0	0.17	50.0	110.0	300.0	625.0	1200.0		(790.3)		D	0.00150
Tested at 700°C (1290°F)															
13709	4430	T T	173	25.0	0.18	0.5	0.7	1.5	2.6	4.5	10.0	52.9	91.0	F	0.4300
12552	4430	T T	173	25.0	0.20	0.2	0.3	0.7	1.3	2.1	5.7	27.0	78.5	F	0.8500
12723	94210	T T	173	25.0	0.13	0.3	1.0	1.9	3.2	6.0	13.0	54.1	100.0	F	0.3900
13160	4240	J T	130	20.0	0.20	0.6	1.5	3.0	6.5	14.0	36.0	140.0	60.0	F	0.1400
13700	4410	T T	130	20.0	0.06	1.2	2.2	5.0	9.0	18.0	40.0	513.4	91.5	F	0.1000
12573	4430	T T	130	20.0	0.05	0.4	2.9	2.1	3.9	6.0	20.0	110.1	67.2	F	0.2500
13149	4406	J T	130	20.0	0.19	0.7	1.4	3.0	4.0	9.0	33.0	182.0	86.0	F	0.1300
12593	94210	T T	130	20.0	0.20	1.2	2.5	4.6	7.0	15.0	39.0	345.2	80.9	D	0.1300
13160	4240	J T	103	15.0	0.14	7.5	5.0	16.0	40.0	100.0	242.0	763.4	40.1	F	0.0100
14490	4410	T L	103	15.0	0.0	6.0	3.0	10.0	23.0	55.0	199.0	2620.2	70.2	F	0.0010
13150	4406	J T	103	15.0	0.15	2.0	5.5	10.5	84.0	140.0	400.0	1400.4	50.9	F	0.0110
12506	4430	T T	93	13.5	0.05	1.9	3.5	10.0	24.0	65.0	243.0	1224.4	28.9	F	0.0150
14106	94210	T L	86	12.5	0.02	6.0	15.0	40.0	105.0	1750.0	5530.0	10640.1	47.7	F	0.00030
13170	4240	J T	69	10.0	0.13	13.0	35.0	190.0	675.0	1400.0	4400.0	(1000.0)		D	0.0010
13094	4410	T L	69	10.0	0.15	10.0	23.0	80.0	245.0	950.0		(570.0)		D	0.00130
13099	4410	T T	69	10.0	0.16	7.0	18.0	82.0	400.0	1300.0		(440.3)		D	0.00110
13010	4430	T T	69	10.0	0.0	8.0	18.0	70.0	330.0	900.0		(440.0)		D	0.00150
13151	4406	J T	69	10.0	0.11	6.0	20.0	90.0	340.0	1500.0		(1497.0)		D	0.00005
12569	94210	T T	69	10.0	0.10	9.0	30.0	107.0	940.0	2650.0		0.0		D	0.00000
12569	94210	T T	69	10.0	0.08	8.0	21.0	85.0	307.0	7800.0	15000.0	29322.2	29.1	F	0.00009
14159	4240	J T	55	8.0	0.02	30.0	100.0	610.0	3200.0			(2992.5)		D	0.00010
13779	4430	T T	55	8.0	0.04	18.0	40.0	240.0	1730.0			(1040.0)		D	0.00017
13000	94210	T T	55	8.0	0.0	11.0	46.0	205.0	1100.0			(1052.0)		D	0.00070
14099	4410	T L	40	7.0	0.07	40.0	110.0	440.0	5500.0			(3560.7)		D	0.00012
Tested at 871°C (1600°F)															
14110	4430	T T	103	15.0	0.0	0.0	0.0	0.0	0.0	0.0	10.7	60.4		F	0.0
14403	94210	T T	103	15.0	0.08	0.1	0.2	0.3	0.7	1.5	4.0	26.3	64.9	F	1.1700
13171	4240	J T	69	10.0	0.13	0.2	0.5	2.0	4.0	9.0	25.0	70.8	33.6	F	0.2000
13004	4410	T T	69	10.0	0.0	0.3	0.6	1.6	3.4	6.5	16.0	76.2	60.9	F	0.3100
12567	4430	T L	69	10.0	0.03	0.1	0.3	0.75	1.8	4.3	11.6	60.5	45.2	F	0.4100
12564	4430	T T	69	10.0	0.10	0.6	1.3	4.5	11.0	23.0	69.0	307.9	41.9	F	0.0700
13152	4406	J T	69	10.0	0.13	0.3	0.8	2.5	6.0	14.0	40.0	161.3	39.9	F	0.1100
12500	94210	T T	69	10.0	0.14	0.5	0.8	2.1	5.1	12.0	30.0	153.6	53.3	F	0.1100
12571	94210	T L	69	10.0	0.08	0.4	0.8	2.8	7.0	17.0	54.0	241.7	63.1	F	0.0000
13172	4240	J T	40	7.0	0.10	1.5	5.0	15.0	32.0	65.0	155.0	443.3	25.8	F	0.0300
12596	4430	T T	40	7.0	0.11	0.5	2.5	20.0	51.5	132.0	320.0	806.2	34.7	F	0.0130
13153	4406	J T	40	7.0	0.10	2.5	8.0	30.0	72.0	150.0	335.0	625.8	21.1	F	0.0120
12595	94210	T T	40	7.0	0.05	1.5	5.0	22.0	120.0	352.0	719.0	1875.1	40.0	F	0.0070
13173	4240	J T	34	5.0	0.07	9.6	25.6	89.0	201.0	430.0	870.0	1652.0	30.6	F	0.0070
13497	4410	T T	34	5.0	0.06	8.0	17.0	51.0	129.0	250.0	520.0	1625.4	42.2	F	0.0070
12575	4430	T L	34	5.0	0.0	2.5	10.0	51.0	176.0	400.0	900.0	2363.0	37.6	F	0.00390
12801	4430	T T	34	5.0	0.07	2.0	9.0	77.0	332.0	720.0	2070.0	3912.0	34.6	F	0.00220
13154	4406	J T	34	5.0	0.07	2.3	23.0	130.0	320.0	650.0	1000.0	2194.7	19.0	F	0.00240
12551	94210	T T	34	5.0	0.0	12.0	40.0	400.0	1000.0	1000.0	1000.0	8506.2	32.0	F	0.00050
12570	94210	T L	34	5.0	0.11	5.2	23.0	230.0	550.0	650.0		(1090.7)	43.1	X	0.00065
13174	4240	J L	20	4.0	0.07	15.5	51.5	205.0	500.0	1000.0	2500.0	(1000.0)		D	0.00160
13097	4410	T L	20	4.0	0.0	22.0	52.0	162.0	350.0	725.0	1750.0	(568.8)		D	0.00260
13347	4410	T T	20	4.0	0.04	3.0	15.0	85.0	220.0	400.0	750.0	2534.4	44.4	F	0.00300
12705	4430	T T	20	4.0	0.04	9.0	28.0	130.0	370.0	700.0	1420.0	3706.2	29.1	F	0.0010
13155	4406	J L	20	4.0	0.04	40.0	132.0	520.0	1190.0	2475.0		(1000.0)		D	0.00070
14095	4240	J T	21	3.0	0.02	50.0	160.0	121.0	3000.0			(3400.0)		D	0.00022
13702	4410	T T	21	3.0	0.0	70.0	170.0	350.0	900.0	1850.0		(1147.4)		D	0.00000
14094	4430	T T	21	3.0	0.0	12.0	53.0	450.0	2100.0			(1909.7)		D	0.00015
14126	4447	T L	20	4.0	0.02	4.0	20.0	135.0	475.0					D	0.0015
14187	4447	T T	20	4.0	0.03	3.2	23.0	150.0	620.0					D	0.001
13817	4455	T T	20	4.0	0.0	10.0	25.0	162.0	800.0					D	0.00075
14000	4392	T L	20	4.0	0.01	3.0	10.0	66.0	130.0					D	0.0045
14063	4392	T T	20	4.0	0.09	5.0	19.0	102.0	295.0					D	0.0026
13941	4455	T L	20	4.0	0.02	15.0	40.0	160.0	460.0					D	0.0015

^aWeld lines of material represented.^bFirst column: tests conducted at T = Stellite Technology Laboratory; and J = Joliet Metallurgical Laboratory. Second column: T = transverse to rolling direction; and L = longitudinal.^cStrain that occurred on loading at start of test.^dRecorded times do not include strain measured on loading.^eNumbers in parentheses indicate running time of test when it was discontinued prematurely.^fF = normal failure; X = test discontinued because of equipment failure at the time recorded under "Time to Rupture;" and D = test discontinued before failure at the time given under "Time to Rupture."

APPENDIX B

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APPENDIX B

Treatment of multiple-heat or heterogeneous data as a single-data population for equation fitting can in some cases be a convenient means to obtain an expression for average behavior of the population as a whole. However, if differences among heats are ignored totally, large inaccuracies and biases can be introduced into the results. This appendix illustrates a few possible simplified examples of problems that can arise in such cases.

Case 1: The most obvious example of biases obtained in multiple-heat analyses occurs when the heats for which data are available are not representative of the behavior of a material in general. For example, if data are available for only a few heats, those heats may be generally stronger or weaker than is truly representative of the material involved. Since the present data for Hastelloy alloy X primarily involve five heats, such a bias may certainly exist here. Unfortunately, there is no analytical solution to lack of data.

Case 2: Even if the heats themselves represent average behavior, lack of homogeneity in the distribution of data can cause errors in fitting. Missing data points, particularly for especially strong or weak heats, are a good example. Such a situation is shown schematically in Fig. B1, where three heats — one strong, one weak, and one average — have been tested. At the lowest stress the test on the strongest heat was discontinued before it reached 1% strain (a common occurrence). The true average behavior is represented by the dashed line in the figure, but a fit in terms of $\log t_{1\%}$ as the dependent variable (which it is) would yield the solid line if a model allowing such curvature were fit. This situation illustrates cases in which model flexibility (such as $\log \sigma$ - $\log t_{1\%}$ curvature) can aggravate biases caused by the data base.

Case 3: Often, data for some heats are available under very limited conditions. For example, at a single high stress many heats may be tested, but only a few heats may be tested over an extended stress range. Such instances can lead to situations such as that shown in Fig. B2.

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Fig. B1. Difficulties Caused by Key Data Missing from a Multiple-Heat Data Set. The dashed line represents the true average; the solid line represents a fit to the data shown.

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Fig. B2. Difficulties Caused When Several Heats are Tested Under Limited Conditions. The dashed line represents the true average; the solid line represents a fit to the data shown.

Case 4: Especially if the data are from different sources, different heats may be tested under very different conditions. For example, if data for heat A (a strong heat) are available only at, say, 649°C, and data for heat B (a weak heat) are available only at, say, 704°C, the heat-to-heat effects and temperature effects will be confounded. It is very difficult, especially without detailed analysis, to discern which part of the difference between the two data sets is caused by the fact that heat A is stronger and which part is caused by temperature.

APPENDIX C

APPENDIX C

The analysis of rupture data for design generally has two goals. One goal involves an attempt to describe actual material behavior. That goal has been dealt with above. The other goal involves setting safe design lower limits on behavior so that rupture of components in service will be precluded. A detailed discussion of such limit-setting procedures is beyond the scope of this paper. However, the regression models developed by the current methods are particularly amenable to limit-setting treatments. Therefore, a brief description of some possible limit-setting procedures will be given here.

Limit-setting procedures can be either statistical or engineering ones. Statistical limits may include several basic types, including confidence limits, prediction limits, and tolerance limits. A general discussion of statistical limits is given in ref. 1. The advantage of these limits is that they are well-defined and have clear, quantitative implications. However, they involve certain assumptions such as that the experimental data obey a certain distribution, usually normal. Violation of these assumptions removes the quantitative meaning of the limits and can in fact make them misleading. Also, these limits are intended primarily for use within the range of the experimental data. Use of such limits for extrapolation beyond the data base (such as is generally necessary with creep data) is dangerous at best.

Engineering limits have the disadvantages that they are somewhat arbitrary in nature and can rely heavily on the judgment of the individual analyst. However, they have the advantage of being flexible enough to yield a "reasonable" estimate of lower limit behavior even in the extrapolated region. They also do not necessarily rely on specific assumptions as to the data distribution. A common method of setting such limits involves subtracting a constant multiple of the standard error of estimate, SEE, from the mean value of stress at a given rupture time or rupture time at a given stress.

Our recent analysis^{2,3} of austenitic stainless multiple-heat data sets suggests an alternative, more quantitative method of setting limits.

The models involving ultimate tensile strength (U) terms will yield approximately average predictions of creep strength when an estimated average value of U is inserted into the model. Similarly estimated maximum and minimum U values will yield estimates of maximum and minimum creep properties.

The choice of the particular method to use depends on factors such as the specific purpose of the analysis, distribution of the available data, etc. Most design codes specify that an additional safety factor (allowable stress reduction) be applied to the "lower limit" value for additional conservatism, whatever method is used. The regression approach described above has the advantage of being easily adaptable to any of the limit-setting techniques discussed above.

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APPENDIX D

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APPENDIX D

COMPARISON OF CURRENT RESULTS WITH OTHER RECENT ANALYSES

Many of the data examined in this report were also recently examined by investigators at AiResearch Corporation in Phoenix (AIRPHX)¹ and at NASA-Lewis Research Center² in support of the BIPS program. These results are compared here with the current results in terms of time to 1% creep strain. Similar comments would apply to analyses for time to rupture.

Unfortunately, the recent analyses performed at AIRPHX¹ and at NASA-Lewis² differed somewhat in their predictions for the seven-year 1% creep strength and rupture strength of Hastelloy alloy X. This evaluation was undertaken with three purposes in mind: to examine the methods and assumptions used in the previous^{1,2} analyses; to compare the techniques and results of the previous analyses to determine which, if either, is more reliable; and, if necessary, to suggest an alternative approach to the two existing analyses.

PREVIOUS RESULTS

The existing AIRPHX¹ and NASA² analyses for time to 1% creep strain of Hastelloy alloy X were evaluated and compared within the framework of the analysis techniques described in the body of this report. Both AIRPHX and NASA workers primarily utilized data from Cabot-Stellite³ investigations (119 data for AIRPHX, 124 for NASA). In addition both used data from seven tests conducted in air at ORNL. The number of Cabot data used in the two investigations differs because five tests were intentionally omitted from the AIRPHX analysis due to apparent inconsistencies in the data for those tests.⁴ Both analyses used data collected at temperatures from 649 to 1093°C (1200–2000°F).

The basic approaches used by both AIRPHX and NASA were also similar. An analytical relationship was sought among the variables stress (σ), temperature (T), and time to the accumulation of 1% creep strain ($t_{1\%}$).

The treatment was essentially identical to that commonly used for the "parametric" analysis of creep-rupture data. Of course, this approach ignored data for other strain levels and thus incorporated less information than was used in the current analysis.

AIRPHX ANALYSIS

The AIRPHX analysis assumed (apparently after some preliminary graphical assessments) a Larson-Miller⁵ parametric fit of the form

$$10^{-3}(T_F + 460)(\log t_{1\%} + 17.9) = A_0 + A_1 \log \sigma + A_2(\log \sigma)^2 + A_3(\log \sigma)^3, \quad (1)$$

where the coefficients A_0 - A_3 were apparently determined by nonlinear regression analysis. The value 17.9 for the "Larson-Miller constant" was arrived at as a compromise value from values determined from data for time to 1, 2, and 5% creep strain and time to rupture. Design lower limits were then obtained by subtracting multiples of the "standard deviation" from the parameter itself.

The AIRPHX results appear to describe the available data for $t_{1\%}$ reasonably well in most cases, but the analysis might be improved in reliability in several ways, enumerated below.

1. The assumption that the Larson-Miller constant is the same for $t_{1\%}$, $t_{2\%}$, $t_{3\%}$, and t_r may facilitate parametric comparison of the criteria, but it introduces an unnecessary bias into the individual results. Such comparisons are probably more meaningful when made directly in terms of time or stress, since the real significance of the parameter is far less clear than that of stress or time.

2. The actual mechanics of the analysis could be greatly simplified by a minor transposition of the terms in Eq. (1), yielding a model of the form

$$\log t_{1\%} = a_0 + \frac{a_1 + a_2 \log \sigma + a_3(\log \sigma)^2 + a_4(\log \sigma)^3}{T_F + 460}. \quad (2)$$

With this model the data can be fitted by linear least squares to optimize predictions for $\log t_{1\%}$, which is the true dependent variable in any case. Design limits can then be set in a more straightforward way by subtracting a multiple of the standard error of estimate from the fit for $\log t_{1\%}$ directly.

3. Examination of the experimental data shows little or no curvature in plots of $\log \sigma$ vs $\log t_{1\%}$. Therefore, the cubic log stress dependence in Eq. (1) does not appear justified. Such a dependence is often used with creep data,⁶ but only if the coefficient in the $[(\log \sigma)^3]$ term is negative. The AIRPHX analysis yielded a positive value for A_3 , forcing their model intrinsically to predict poor behavior upon extrapolation. This behavior will be demonstrated in this report.

4. The choice of data base for the analysis appears questionable. In general, treating scatter bands of multiple-heat data without considering the behavior of individual heats can be very dangerous. It is not clear that AIRPHX performed any extensive single-heat evaluations. For example, our analyses showed the ORNL data to differ somewhat from the Cabot data, perhaps because the product forms differed. To avoid bias in the results, we excluded ORNL data from our analysis. Also, data to 1093°C (2000°F) are probably irrelevant for BIPS applications. The rationale for including such high-temperature data was probably to obtain lower stress points on the Larson-Miller "master curve." However, realize that even though a time-temperature parameter fits the available data, that same parameter may not yield accurate extrapolations. Parametric prediction of low-stress, low-temperature data by higher temperature data at those stresses is an extrapolation and should only be done with extreme caution. Figure D1 illustrates this situation schematically. Thus, even though a point may lie within the range of the "master curve," it can still represent an extrapolation and should be treated accordingly.

5. Finally, it may be wise to consider a wider range of models than those evaluated by AIRPHX. Time-temperature parameters represent only a small subset of possible models, and any given parameter imposes certain well-defined assumptions on the final results. (Larke and Inglis⁷ analyzed the restrictions imposed by various commonly used parameters.)

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Fig. D1. Region of Interpolation Versus Extrapolation for Creep Data. Any point within the boxed region lies within the parametric "master curve," while only points within the shaded region lie within the range of experimental data.

NASA ANALYSIS

The model employed by NASA was of the form

$$1/t_{1X} = k(\sinh \alpha\sigma/E)^n (e^{-Q/RT} + ce^{-Q_2/RT}) , \quad (3)$$

where the constants k , α , n , Q , c , and Q_2 are determined from fits to the data. The constants α , Q , c , and Q_2 were assumed to be the same for t_{1X} , t_{2X} , t_{5X} , and t_T . This model is based on several theoretical considerations.⁸⁻¹⁰

Although the approach proposed by NASA is more clearly related to physical assumptions about the creep behavior of Hastelloy alloy X than is the AIRPHX analysis, it is still essentially a parametric analysis of

the same data. As such, objections 4 and 5 above still hold, although single-heat analysis was performed.¹¹ The assumed invariance in α , Q , c , and Q_2 can cause problems similar to those mentioned in objection 1 above. Two additional comments also apply.

First, the rationale for choice of the NASA model seems to have been largely physical. In reality, the physical assumptions implied in the model may not be particularly realistic for a multiple-heat data base of a complex alloy. Interestingly, the NASA analysis appeared to identify the existence of different creep mechanisms above and below 871°C (1600°F). Thus, the use of data collected at temperatures above this point is again questioned.

Second, the NASA equation is again a complicated, nonlinear (in a regression sense) model. Similar results could be obtained much more easily with a model of the form

$$\log t_{1\%} = b_0 + b_1 \log \sigma + b_2 \sigma + b_3/T + b_4/T . \quad (4)$$

The equation used [Eq. (3)] is probably more complex than the data require or justify.

CURRENT RESULTS

The predictions from the current analysis for multiple-heat data are compared in Fig. D2 with experimental data and with the predictions obtained by AIRPHX and NASA. Figure D3 estimates the seven-year 1% creep strengths from each of the above analyses. Also included in the table are estimated strengths for each data set obtained by simple graphical straight line fitting and extrapolation of the $\log \sigma$ - $\log t_{1\%}$ isotherms.

The above data and figures reveal several trends. First, the NASA model appears, in the judgment of this author, overconservative at long times, perhaps due to the influence of the high-temperature data on predictions from the model form chosen. Second, the inherently unstable nature of the AIRPHX model can be noted in the 871°C results in Fig. D2. The ORNL model appears to adequately reflect trends in both the single-heat

Fig. D2. Comparison of Experimental Data for Time to 1% Creep Strain for Hastelloy Alloy X with Predictions from ORNL, AIRPHX, and NASA Multiple-Heat Fits.

Fig. D3. Comparison of Estimated Average Seven-Year 1% Creep Strengths for Hastelloy Alloy X from ORNL, AIRPHX, and NASA Analyses.

and multiple-heat data sets. Still, the variations in strength levels estimated by various techniques — especially of the single-heat results — illustrate the uncertainties involved in estimating seven-year 1% creep strength from the available data. Use of average -2 SEE "design" predictions appears adequate for these data, although the variations in strength among the various heats are fairly large, indicating that the problem of heat-to-heat variations in properties should be a serious concern for BIPS design. Figures D3 and D4 compare the predictions from the analysis with those of AIRPHX and NASA. The variations among the different predictions can be judged from the figures. Essentially, the ORNL results resemble those of AIRPHX, while the NASA stress predictions are somewhat lower. Given the discussion above, the current results appear to be preferable.

Fig. D4. Comparison of Estimated Design Seven-Year 1% Creep Strengths for Hastelloy Alloy X from ORNL, AIRPHX, and NASA Analyses.

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