A STUDY OF TERTIARY CREEP INSTABILITY IN SEVERAL ELEVATED-TEMPERATURE STRUCTURAL MATERIALS*

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ABSTRACT

In the current investigation data for a number of common elevated temperature structural materials have been analyzed to yield mathematical predictions for the time and strain to tertiary creep at various rupture lives and temperatures. Materials examined include types 304 and 316 stainless steel, 2 1/4 Cr–1 Mo steel, alloy 800H, alloy 718, Hastelloy alloy X, and ERNiCr–3 weld metal. Data were typically examined over a range of creep temperatures for rupture lives ranging from less than 100 to greater than 10,000 hours. Within a given material, trends in these quantities can be consistently described, but it is difficult to directly relate the onset of tertiary creep to failure-inducing instabilities. For example, a series of discontinued tests for alloy 718 at 649 and 620°C showed that the material fails by intergranular cracking but that no significant intergranular cracking occurs until well after the onset of tertiary creep.

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DISCLAIMER

INTRODUCTION

Current design rules[1] for elevated-temperature nuclear systems require consideration of the time to the onset of tertiary creep as one factor in the determination of design allowable stress intensity limits. Reasons for this criterion include the prevention of metallurgical or geometric instabilities that might lead to failure or to excessive strain during service. Unfortunately, the manifestations of tertiary creep can vary widely from material to material, and the actual physical meaning of this criterion is somewhat unclear.

In recent investigations[2–12] creep data for a number of common elevated-temperature structural materials have been examined in order to describe the manifestations of tertiary creep in various systems. Materials examined include types 304 and 316 austenitic stainless steels, ferritic 2 1/4 Cr—1 Mo steel, nickel base alloy 718, Hastelloy alloy X, class ERNiCr—3 weld metal, and the high nickel austenitic alloy 800H. Features studied include the strain incurred before the onset of tertiary creep and its relationship to the total creep strain to rupture and the time to tertiary creep and its relation to the time to rupture. This paper seeks to summarize and interpret the results of those investigations.

In addition, creep tests for alloy 718 have been discontinued at various points from about 20% of the rupture life to the actual rupture point. Phenomenological examination of experimental data and metallographic examination of specimens from the discontinued tests have been used to relate the occurrence of tertiary creep to micorstructural changes in the material. It is shown that the observed onset of tertiary creep in terms of increasing creep rates does not necessarily imply instability or impending failure.

DATA USED

Data examined for type 304 stainless steel were derived primarily from tests conducted at Oak Ridge National Laboratory (ORNL) on 20 heats of annealed material. References 4 and 13 describe the behavior of those heats. Data for type 316 stainless steel were taken primarily from two heats[14,15] with additional recent data having been generated at ORNL. Data for several heats of alloy 800H were taken from a variety of sources, as described elsewhere.[5,6] Information for 2 1/4 Cr—1 Mo steel included annealed and isothermally annealed material with or without subsequent simulated postweld heat treatments. Air-melted, electroslag remelted (ESR), and vacuum arc remelted (VAR) materials were considered. Data were taken from Refs. 16—19 and were analyzed in detail previously.[7,8] Data for alloy 718 were obtained from recent ORNL test programs[9] for three different heats of material. Data for Hastelloy alloy X were taken from Ref. 20 and are analyzed in Ref. 11. Reference 12 presents detailed analysis of data[21,22] for class ERNiCr—3 weld metal (commonly known as Inconel 82).

DEFINING THE ONSET OF TERTIARY CREEP

One factor that complicates analysis for data describing the onset of tertiary creep is the problem of defining the onset point for individual creep strain-time curves. For a classical curve as shown in Fig. 1, the times t_2 and t_{ss} are both commonly used measures of the time to the onset of tertiary creep, with e_2 and e_{ss} being the corresponding creep strains The definition problem becomes much more complicated as the creep curve shape becomes less classical.

The creep strain-time behavior of 2 1/4 Cr—1 Mo steel can be quite complicated. This material strongly tends toward nonclassical creep behavior (Fig. 2). This behavior is discussed in more detail by Klueh,[16,17] who points out that the increase in creep rates at the end of the first steady-stage does not represent the onset of tertiary creep (as one might wrongly be led to believe). Rather, it probably represents a metallurgical instability within the material. The creep rate is controlled by different mechanisms in the two steady-state stages.

The existence of two separate steady-state stages appears to be a dominant feature of the behavior of this material and should be accounted for by any predictive model. For the nonclassical curve (Fig. 2) we have called[7,8] the first primary and steady-state portions "Type I" creep and the second linear portion "Type II" creep. The stress-temperature conditions which produce classical or nonclassical creep vary with heat and heat treatment.

If the transition from Type I to Type II creep is wrongly assumed to be the onset of tertiary creep, then 2 1/4 Cr—1 Mo steel appears to display early onset of tertiary creep. However, for the classical curves available, the time to the onset of tertiary creep is always about half the rupture life, t_r. For the nonclassical curves the end of the second linear stage is also about half the rupture life. For this reason it is thought that the true onset of tertiary creep begins at the end of this second linear stage. The end of the first linear stage represents merely a mechanism change. Onset of tertiary creep was differentiated from the end of Type I creep through the classification procedure described elsewhere.[7,8]

Hastelloy alloy X often displays nonclassical creep curves similar to those noted for 2 1/4 Cr—1 Mo steel. At present no metallurgical explanation has been developed for this behavior. However, it was concluded[11] that the end of the second linear stage was appropriately called the onset of tertiary creep for this material as well.

Both alloys 800H and 718 show tendencies toward unusually early onset of tertiary creep. As shown in Fig. 3 the creep curves for such tests often do not display a clear region of linear creep at all, making measurement of the time to tertiary creep particularly difficult. In the current analysis t_2 and t_{ss} for alloy 800H were determined as shown in Fig. 1. For alloy 718 only t_{ss} was determined since we felt that t_{ss} could be more consistently determined than t_2 from the curves for that material. Types 304 and 316 stainless steel usually display fairly classical behavior, although unusual phenomena do occur (Fig. 4). The behavior of ERNiCr—3 weld metal varies with temperature, as will be described below.

Noting the above differences in behavior from material to material, values for t_{ss}, t_2, e_{ss}, and e_2 were determined from available creep curves. These values were next examined to identify trends in behavior for the various methods.

PHENOMENOLOGY OF TERTIARY CREEP

TIME TO TERTIARY CREEP

Probably the most widely used[2,3,10,15,23–25] method for analysis of tertiary creep data is the examination of relationships between the time to tertiary creep and the corresponding time to rupture, t_r. All of the above materials have been shown[5,6,9,10,13,16] to obey a relationship of the form

$$t_3 = At_r^{\beta} , \qquad (1)$$

where t_3 may be t_{ss} or t_2 and where A and β are functions of material only, except for alloy 800H where A is a function of temperature as well and ERNiCr—3 where both A and β vary with temperature.

For ERNiCr—3 in the temperature range from 454 to 566°C. there was essentially no tertiary creep portion, with rupture occurring very quickly after deviation from linear creep. In this regime t_2 and t_{ss} are essentially equal, and either can be described by

$$t_{ss} = t_2 = 0.98t_r . \qquad (2)$$

Data for t_{ss} in the range 677—732°C showed curvature when plotted log-log against t_r (Fig. 5). This phenomenon can be attributed to the shapes of the creep curves in this range. As times become longer the creep strain to tertiary creep becomes smaller, and the constant 0.2% offset becomes relatively larger. As a result t_{ss} becomes a larger and larger fraction of t_r. When viewed in terms of t_2, these plots are again essentially linear, with the behavior being described by

$$t_2 = 0.035t_r^{1.20} , \qquad (3)$$

as shown in Fig. 6. This equation will eventually yield $t_2 > t_r$ at very long times. For the longest term available data $t_2 \approx 0.2t_r$. Thus we (arbitrarily) recommend predicting t_2 by Eq. (3) or by $t_2 = 0.6t_r$, whichever is smaller, since $0.6t_r$ falls midway between $0.2t_r$ and t_r, the range in which t_2 must lie. The behavior seen in Fig. 5 indicates that perhaps a variable offset would be more appropriate than the fixed (arbitrary) value of 0.2%. At 621°C t_2 was given by approximately $0.85t_r$, showing behavior intermdeiate to the above two temperature regimes. Figures 7 and 8 illustrate the applicability of Eq. (1) to other materials, while Table 1 summarizes the values of A and β.

It should be noted that the time to tertiary creep can be treated directly as a function of stress and temperature.[4,5,9,10] Such analysis can be performed using techniques similar to those commonly used for stress-rupture data.[26,27]

STRAIN TO TERTIARY CREEP

Techniques used by the authors for estimation of the strain incurred before the onset of tertiary creep are described in detail elsewhere.[2,3,28] The method is essentially that proposed by Smith[29] and expanded by Goldhoff[30] for strain to rupture. Using e_{ss} as an example (the treatment of e_2 is analogous), the average creep rate to tertiary creep $\dot{e}_{ss}$ is defined by

$$\dot{e}_{ss} = e_{ss}/t_{ss} \quad . \tag{4}$$

Although the scatter in e_{ss} and e_2 is often too great to permit a meaningful direct analysis, the quantities $\dot{e}_{ss}$ and $\dot{e}_2$ usually exhibit far less scatter. They can then be treated directly by parametric, regression, or other techniques. We have generally obtained optimum results with the relationships

$$\dot{e}_3 = B\dot{e}_m^{\alpha} \tag{5}$$

or

$$\dot{e}_3 = Dt_r^{-\gamma} , \tag{6}$$

where $\dot{e}_3$ is $\dot{e}_{ss}$ or e_2 and where B, α, D, and γ are material constants. Figures 9 and 10 illustrate tha applicability of Eq. (5) , while Table 2 summarizes the results for various materials.

In some cases parameter D in Eq. (6) was found to be a definite function of temperature. Table 3 summarizes the correlations obtained for some of the materials by use of an equation of the form

$$\dot{e}_3 = D_0 e^{-Q/RT} t_r^{-\gamma_0} , \tag{7}$$

where D_0, γ_0, and Q are temperature-independent material constants, R is the gas constant (8.31 J/mole), and T is the temperature in K. As illustrated in Fig. 11 the fits are again quite good.

Data for $\dot{e}_3$ can also be examined directly as a function of stress and temperature. Such direct analyses for alloy 800H (for $\dot{e}_2$ and t_2) yields a final equation for e_2 as

$$e_2 = 7.238 - 11,300/T + \frac{2,150}{T}\log\sigma , \tag{8}$$

where

T = temperature (K) and

σ = stress (MPa).

Whatever methods are used, estimation of t_3 and $\dot{e}_3$ then yields e_3 as as the product of the two. Figures 12 and 13 compare typical predictions with data for creep strain to tertiary creep.

STRAIN TO RUPTURE

The creep strain to rupture, e_r, can be treated by methods analogous to those used above for e_3. In many cases the only available measure of rupture ductility is in terms of total strain (creep plus instantaneous plastic strain on loading), e_t. Fortunately, the two quantities, e_r and e_t, can be treated similarly. Good results have generally been obtained by relating the average strain rates $\dot{e}_r = e_r/t_r$ and $\dot{e}_t = e_t/t_r$ by

$$\dot{e}_r = Et_r^{-\delta} \tag{9}$$

$$\dot{e}_t = Gt_r^{-\varepsilon} \tag{10}$$

where the material constants E, δ, G, and ε may all be functions of temperature. This temperature dependence can sometimes be described in the form

$$\dot{e}_r = E_0 e^{-Q/RT} t_r^{-\delta_0} \ , \tag{11}$$

where R is the gas constant, E_0, Q, and δ_0 are the material constants, and T is the absolute temperature. On the other hand, for 2 1/4 Cr—1 Mo steel the temperature dependence is negligible. The quantity $\dot{e}_r$ can also be directly treated as a function of stress and temperature (as can $\dot{e}_3$). Figures 14 and 15 illustrate the relationship between average strain rate to rupture and rupture life, while Fig. 16 compares typical predictions with data for strain to rupture. Optimum results for the current materials were obtained from the following equations:

304 Stainless

(482–816°C) $\dot{e}_r = 22{,}578e^{-50{,}559/RT}t_r^{-1.806}$ (12)

316 Stainless

(538–760°C) $\dot{e}_r = 1.061 \times 10^6 e^{-79{,}776/RT}t_r^{-1.029}$ (13)

2 1/4 Cr–1 Mo

(454–566°C) $\dot{e}_t = 27t_r^{-1.0}$ (14)

Alloy 718

(538–760°C) $\dot{e}_r = 136{,}700e^{-72{,}210/RT}t_r^{-1.0}$ (15)

Alloy 800H

(538–871°C) $\log\dot{e}_t = 25.18 - 41{,}200/T + \frac{7{,}410}{T}\log\sigma$ (16)

Hastelloy Alloy X

(649–760°C) $\log\dot{e}_t = 24.02 - 39{,}080/T - 2.666\log\sigma + \frac{9{,}140}{T}\log\sigma$ (17)

ERNiCr–3 Weld Metal

(454°C) $\dot{e}_r = 38t_r^{-1.0}$ (18)

(510–566°C) $\dot{e}_r = 0.0042e^{71{,}020/RT}t_r^{-1.3}$ (19)

(621–732°C) $\dot{e}_r = 54{,}690e^{-50{,}110/RT}t_r^{-1.3}$ (20)

DISCUSSION

TRENDS IN BEHAVIOR

Table 4 summarizes predicted trends in e_3, t_3, e_r (or e_t), and e_3/e_r as functions of rupture life at different temperatures. As discussed elsewhere,[28] trends in creep ductility data are often more clearly discernable in terms of stress than time, but time is a convenient index from which to draw inferences about the importance of various trends for design applications.

Trends clearly vary from material to material. The ratio t_3/t_r may be constant, may increase or decrease with time, and may decrease with temperature. The predicted quantity e_3 always decreases with time except for 2 1/4 Cr—1 Mo steel, where it remains constant within the data used here. For 304 and 316 stainless steels e_3 increases with temperature, while for 2 1/4 Cr—1Mo steel, alloy 718, alloy 800H, and Hastelloy alloy X, e_3 shows no temperature dependence within the range studied (Table II). Trends in e_3 (and e_t) for ERNiCr—3 weld metal with temperature are complex. The strain to rupture (e_r or e_t) decreases with time except for 2 1/4 Cr—1 Mo steel and alloy 718 (and ERNiCr—3 at 454°C). Rising temperature may yield increasing values for e_r or e_t (304 or 316 stainless, alloy 718, alloy 800H, ERNiCr—3 at some temperatures), decreasing values for e_r or e_t (Hastelloy alloy X, ERNiCr—3 at some temperatures), or constant values (2 1/4 Cr—1 Mo steel). The ratio of e_3/e_r (or e_3/e_t) can go up or down with time or temperature.

Examination of Table 4 gives a good idea of expected typical trends for the materials included in this study, and those trends will not be reiterated in the text. However, of particular interest are the extremely low values of e_3/e_r for some of the materials, indicating a substantial capacity for straining even beyond the onset of tertiary creep.

It should be noted that the values in Table 4 are intended to show typical trends in behavior. Creep ductility data show considerable scatter, so actual values may vary somewhat from those given in the table. Effects such as heat-to-heat variations in behavior can also be important, but the values in the table give a good idea of typical properties based on available data.

TERTIARY CREEP AS A DESIGN CRITERION

Current elevated temperature design rules[1] require consideration of the stress to cause (within the design life) rupture, onset of tertiary creep, or 1% total through-thickness strain.

The actual occurrence of creep rupture is an obvious design criterion in that it is directly related to failure. In some cases the onset of tertiary creep can also be related to failure. Penny and Mariott[31] express the view that creep rupture as an instability is more related to the conditions preceding and causing the onset of the instability than to the conditions after the instability occurs. Therefore, the onset of tertiary creep, if viewed as the beginning of failure, can have fundamental significance. Grant et al.[32,33] support this view, finding that the onset of tertiary creep (especially in terms of ductility) can be more fundamental than rupture itself.

Presumably, onset of tertiary creep as a design criterion is meant to prevent excessive internal cracking and voidage and to prevent gross plastic instability. Garofalo[34] indicated that the onset of tertiary creep can be related to several factors, the most commonly mentioned of which include:

1. increasing stress under constant load (result of decrease in cross-section area caused by uniform deformation),
2. increasing stress due to necking (plastic instability),
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 several factors could be responsible for tertiary creep. The exact mechanism and meaning of tertiary creep appear to be complex and probably vary from material to material. It is not clear that the onset of tertiary creep can be directly related to cracking or plastic instability, especially when some uncertainty exists concerning the very definition of the onset of tertiary creep even for a given creep test.

Using the plastic instability criterion proposed by Hart[35] and Burke and Nix[36] showed that plastic instability can be expected to begin in the primary stage of creep. Interestingly, they also found that constant load creep tests (with their steadily increasing stresses) are not inherently more unstable than constant stress tests. No direct relationship was noted between onset of tertiary creep and gross plastic instability.

Söderberg[37] studied the tertiary creep behavior of a 20 Cr—30 Ni steel, finding that the increasing creep rates in constant load creep tests for that material could often be accounted for merely by considering the increasing stress due to decreasing specimen cross-section with straining. On the other hand some grain boundary cracks developed during tertiary creep at low stresses, and both Soderberg[37] and Burke and Nix[36] observed tertiary creep even in constant stress tests.

One study[38] for type 316 stainless steel did reveal a correlation between failure of biaxial tube specimens and onset of tertiary creep in uniaxial specimens. Failure in the tubes was defined by loss in pressure due to crack propagation through the wall. However, the tube walls were quite thin, being less than four grains in average thickness. For the data used here 2 1/4 Cr—1 Mo steel generally does not show grain boundary cracks even at rupture.[39] The failure is typically pure transgranular in nature, which may account for the relative time and temperature independence of the strain to tertiary creep and to rupture. At longer times the failure may become intergranular, however.

From Table 4 it can be seen that alloys 718 and 800H can both display very low strains to tertiary creep. Both of these materials typically display intergranular creep rupture, which could be related to the low values for e_3 if cracking is indeed related directly to the onset of tertiary creep. To investigate this possibility we studied the creep behavior of alloy 718 in more detail. A series of tests was conducted at 620 MPa (90 ksi) and 649°C (1200°F). One test ruptured in 869 hr. Other tests were loaded then discontinued and unloaded after about 0%, 20%, 40%, 60%, and 80% of the life from the rupture test. These points are related to the creep curve at this temperature in

Fig. 17. As shown in Fig. 18 the ruptured test showed extensive grain boundary cracking. None of the other tests showed significant amounts of such cracking, even though the creep rate increases steadily with time beyond about 20% of the life. These increasing rates are not expected to be due to changes in specimen cross-section, since total strains are quite small, at least in the initial portion of the increasing creep rates. No obvious microstructural changes that might be related to these increasing rates were observed through optical or scanning electron microscopy techniques. However, there is evidence that the creep strength of this material is strongly dependent on a finely dispersed γ' phase in the matrix. Transmission electron microscopy studies that might reveal any changes in the size and distribution of γ' particles have not been performed. However, γ' hardening can often be related to hardness values.[41] Room-temperature micro-hardness measurements (Table 5) revealed no discernible relationship between hardness decreases and increasing creep rates.

Phenomenologically there is additional evidence from available stress-relaxation data for alloy 718 that the increasing creep rates at least in early tertiary creep are not a manifestation of instability or impending failure. When a single specimen is loaded, allowed to relax, unloaded, reloaded, relaxed again, and so on, the material displays gradual net softening or increasing relaxation rates from cycle to cycle.[42] As shown in Fig. 19 this softening can be well-simulated by use of a creep equation[9] that includes increasing creep rates in conjunction with the hypothesis of strain hardening.[43] This agreement suggests that the increasing creep rates merely indicate a state (for whatever reason) of decreasing strain hardening, not a state of impending failure.

In short there is little direct evidence for the materials considered here that the onset of tertiary creep is a direct manifestation of excessive internal cracking or gross plastic instability. A second reason for the use of tertiary creep as a design criterion is purely analytical. Many of the currently used creep laws[4-8,12] for design analysis are valid only in the primary and secondary stages of creep. There is also some question concerning possible numerical difficulties that might be encountered by implementing increasing creep rates in standard finite-element computer programs. Limited results[44] from the CREEP-PLAST[45,46] computer code indicated that these increasing rates could increase the analysis cost but appeared to present no insurmountable difficulties. Moreover, recently developed creep laws[9,11] for alloy 718 and for Hastelloy alloy X include regions of increasing creep rate in their formulation.

SUMMARY

The time to the onset of tertiary creep is currently considered as one criterion for setting allowable stresses for elevated temperature nuclear service. Ostensibly this criterion is meant to provide assurance against excessive cracking and void formation and to prevent gross plastic instability. It also serves to prevent possible analytical difficulties with the treatment of increasing creep rates in standard finite element analysis codes. Examination of data for several common elevated temperature structural materials shows that the phenomenological manifestations of tertiary creep can vary significantly from material to material or with stress and temperature for a given material. Indeed, variations in creep curve shape can make it very difficult to establish a consistent definition of onset of tertiary creep even for individual creep curves. Moreover, we have found little evidence of a direct relationship between onset of tertiary creep and cracking or plastic instability, and current design computer codes now appear capable of adequately handling increasing creep rates. On the other hand, rupture as a design criterion does not prevent the unstable conditions leading to rupture. The exact meaning of the tertiary creep criterion is not clear in light of these findings. However, the criterion is probably conservative. While instabilities may not occur until well after the onset of increasing creep rates, they are not likely to occur before the increasing rates.

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Table 1. Relationship between time to tertiary creep and rupture life for various materials

Material	Criterion	Number of data	A	β	Temperature range of data (°C)
304 stainless	t_{ss}	277	0.752	0.977	482–816
304 stainless	t_2	233	0.685	0.968	538–649
316 stainless	t_2	183	0.526	1.004	538–816
2 1/4 Cr–1 Mo	t_{ss}	124	0.54	1.0	454–566
alloy 718	t_{ss}	52	0.392	1.04	538–760
alloy 800H	t_{ss}	50	*a*	0.996	538–871
alloy 800H	t_2	50	*b*	0.940	538–871
Hastelloy X	t_{ss}	33	0.40	1.0	649–871
ERNiCr–3	t_{ss} or t_2	310	0.98	1.0	454–566
ERNiCr–3	t_2	8	0.85	1.0	621
ERNiCr–3	t_2	15	0.035	1.2	677–732

[a] Above $T = 866K$, $A = 0.000628e^{6108/T}$;
$T \leq 866K$, $A = 0.726$

[b] Above $T = 866K$, $A = 0.00135e^{5480/T}$;
for $T \leq 866K$, $A = 0.759$

Table 2. Relationship between average creep rate to tertiary creep and minimum creep rate for various materials

Material	Number of data	B	α	Temperature range of data (°C)
304 stainless	138	1.110	0.974	482–816
316 stainless[a]	120	1.602	0.995	593–816
316 stainless	38	1.092	0.991	538–760
2 1/4 Cr–1 Mo[b]	18	1.5	1.0	454–566
2 1/4 Cr–1 Mo[c]	15	0.8	1.0	510–566
alloy 718[d]	52	2.95	1.18	538–760
alloy 800H	50	0.8	0.839	538–871
Hastelloy X[d]	33	60	1.2	649–760
ERNiCr–3	30	0.84	0.75	454–566
ERNiCr–3	23	1.5	1.0	621–732

[a]Data represented average total strain (creep + plastic + elastic) rate to tertiary creep.

[b]From classical creep curves

[c]From nonclassical creep curves

[d]$\dot{e}_3$ was related to t_r rather than to $\dot{e}_m$ [Eq. (4)]

Table 3. Relationship among average creep rate to tertiary creep, rupture life, and temperature[a]

Material	Number of data	D_0	Q	γ_0	Temperature range of data (°C)
304 stainless	139	41,725	58,809	1.088	482–816
304 stainless[b]	120	132.7	10,063	1.072	593–816
316 stainless	38	205,447	70,358	1.064	538–760
alloy 718	53	2.95	0[c]	1.18	538–760
Hastelloy X	33	60	0[c]	1.2	649–760

[a]Average creep rate to tertiary creep, $\dot{e}_3$, has been expressed as a function of rupture life (hr), t_r, and temperature (K), T, by $\dot{e}_3 = D_0 e^{-Q/RT} t_r^{-\gamma_0}$, where R is the gas constant, 8.31 $J\ mole^{-1} K^{-1}$.

[b]Data represented average total strain (creep + plastic + elastic) rate to tertiary creep.

[c]No temperature dependence in relationship.

Table 4. Summary of predicted behavior for various rupture lives and temperatures

Temperature (°C)	Time to tertiary creep (hr)			Strain to tertiary creep (%)			Strain to rupture (%)			Ratio of strain to tertiary creep to strain to rupture		
	Rupture life (hr)											
	10^1	10^3	10^5	10^1	10^3	10^5	10^1	10^3	10^5	10^1	10^3	10^5
						304 Stainless Steel						
538	7.1[a]	640	58,000	3.9[a]	2.4	1.4	10.3[b]	7.1	4.9	0.38	0.34	0.28
649	7.1	640	58,000	11.3	6.8	4.1	25.5	17.6	12.1	0.44	0.38	0.34
760	7.1	640	58,000	25.7	15.4	9.2	51.8	35.7	24.6	0.50	0.43	0.37
						316 Stainless Steel						
538	5.3[c]	540	55,000	2.8[c]	2.1	1.6	6.3[b]	4.2	2.8	0.44	0.50	0.57
649	5.3	540	55,000	9.7	7.3	5.6	26.2	17.6	11.8	0.37	0.41	0.47
760	5.3	540	55,000	25.9	19.7	14.9	80.1	53.9	36.3	0.32	0.36	0.41
						2 1/4 Cr–1 Mo Steel						
454	5.4[a]	540	54,000	2.5–10.0[a]	2.5–10.0	2.5–10.0	27[d]	27	27	0.092–0.37	0.092–0.37	0.092–0.37
510	5.4	540	54,000	2.5–10.0	2.5–10.0	2.5–10.0	27	27	27	0.092–0.37	0.092–0.37	0.092–0.37
566	5.4	540	54,000	2.5–10.0	2.5–10.0	2.5–10.0	27	27	27	0.092–0.37	0.092–0.37	0.092–0.37
						Alloy 718						
538	4.3[a]	520	57,000	1.0[a]	0.64	0.43	3.0[b]	3.0	3.0	0.33	0.21	0.14
649	4.3	520	57,000	1.0	0.64	0.43	11.0	11.0	11.0	0.091	0.058	0.039
760	4.3	520	57,000	1.0	0.64	0.43	30.4	30.4	30.4	0.033	0.021	0.014
						Alloy 800H						
538	6.6[c]	500	38,000	6.5[c]	1.5	0.32	18.1[d]	9.1	4.5	0.36	0.16	0.071
649	4.5	340	26,000	6.7	1.4	0.30	36.9	18.1	9.0	0.18	0.077	0.033
760	2.4	180	14,000	6.4	1.4	0.31	63.6	31.2	15.3	0.10	0.045	0.020
						Hastelloy Alloy X						
649	4.0[a]	400	40,000	15.1[a]	6.0	2.4	*e*	66.4[d]	45.8	*e*	0.090	0.052
760	4.0	400	40,000	15.1	6.0	2.4	*e*	56.5	33.4	*e*	0.11	0.072
871	4.0	400	40,000	15.1	6.0	2.4	75.4	36.1	17.2	0.20	0.17	0.14
						ERNiCr–3 Weld Metal						
454	9.8[c]	980	98,000	7.3[c]	3.3	1.5	38[d]	38	38	0.19	0.087	0.039
510	9.8	980	98,000	7.3	3.3	1.5	116	29.0	7.3	0.063	0.11	0.20
566	9.8	980	98,000	7.3	3.3	1.5	55.9	14.0	3.5	0.13	0.24	0.43
621	8.5	850	85,000	20.0	5.6	0.08	32.2	8.1	2.0	0.63	0.72	0.50
677	0.5	140	35,000	1.7	0.56	0.35	48.0	12.0	3.0	0.040	0.063	0.18
732	0.5	140	35,000	1.2	0.30	0.05	67.9	17.1	4.3	0.021	0.029	0.058

[a] Values are t_{ss} and e_{ss}

[b] Values are e_r

[c] Values are t_2 and e_2

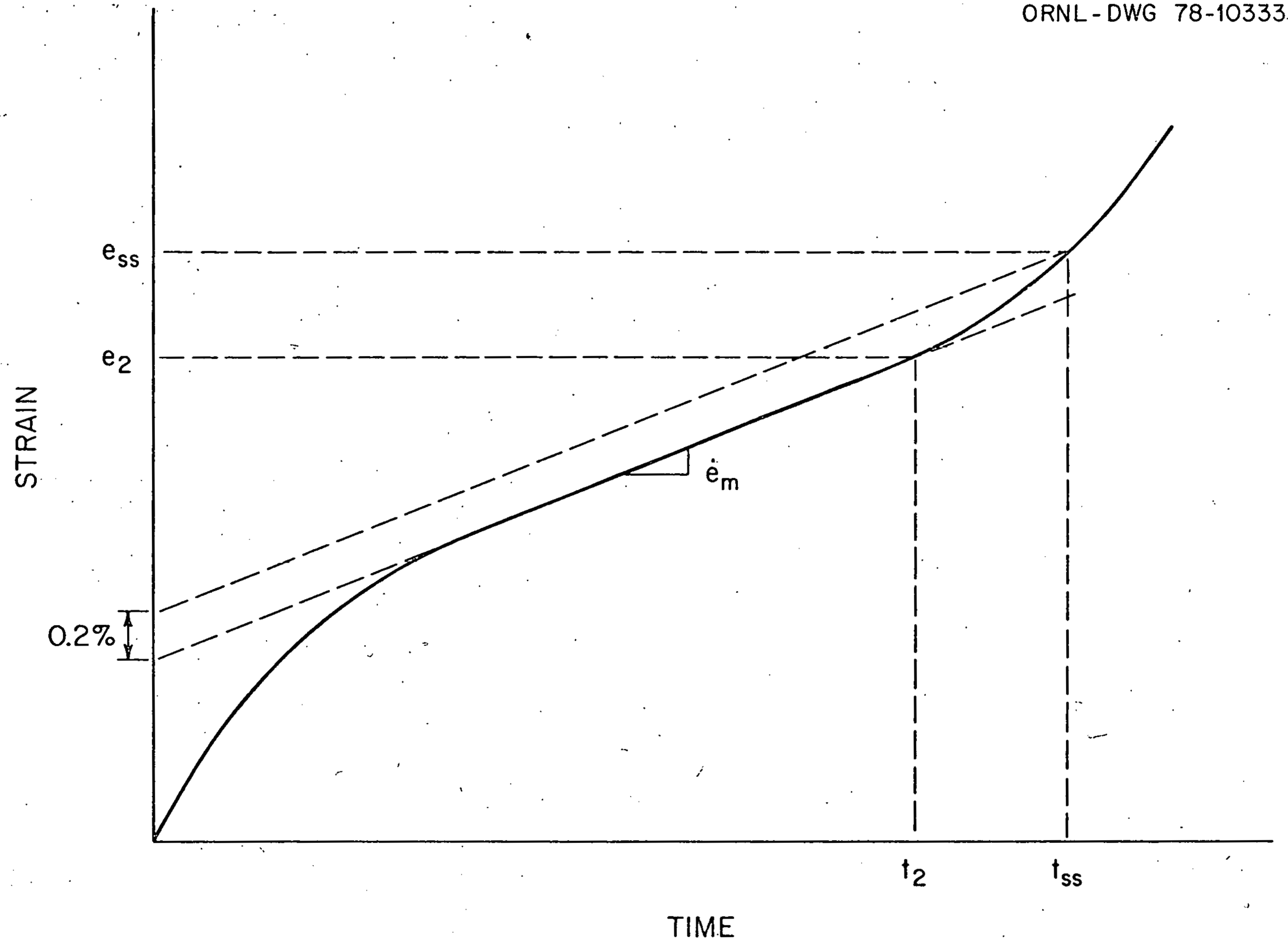
[d] Values are e_t

[e] Predictions not valid in this range

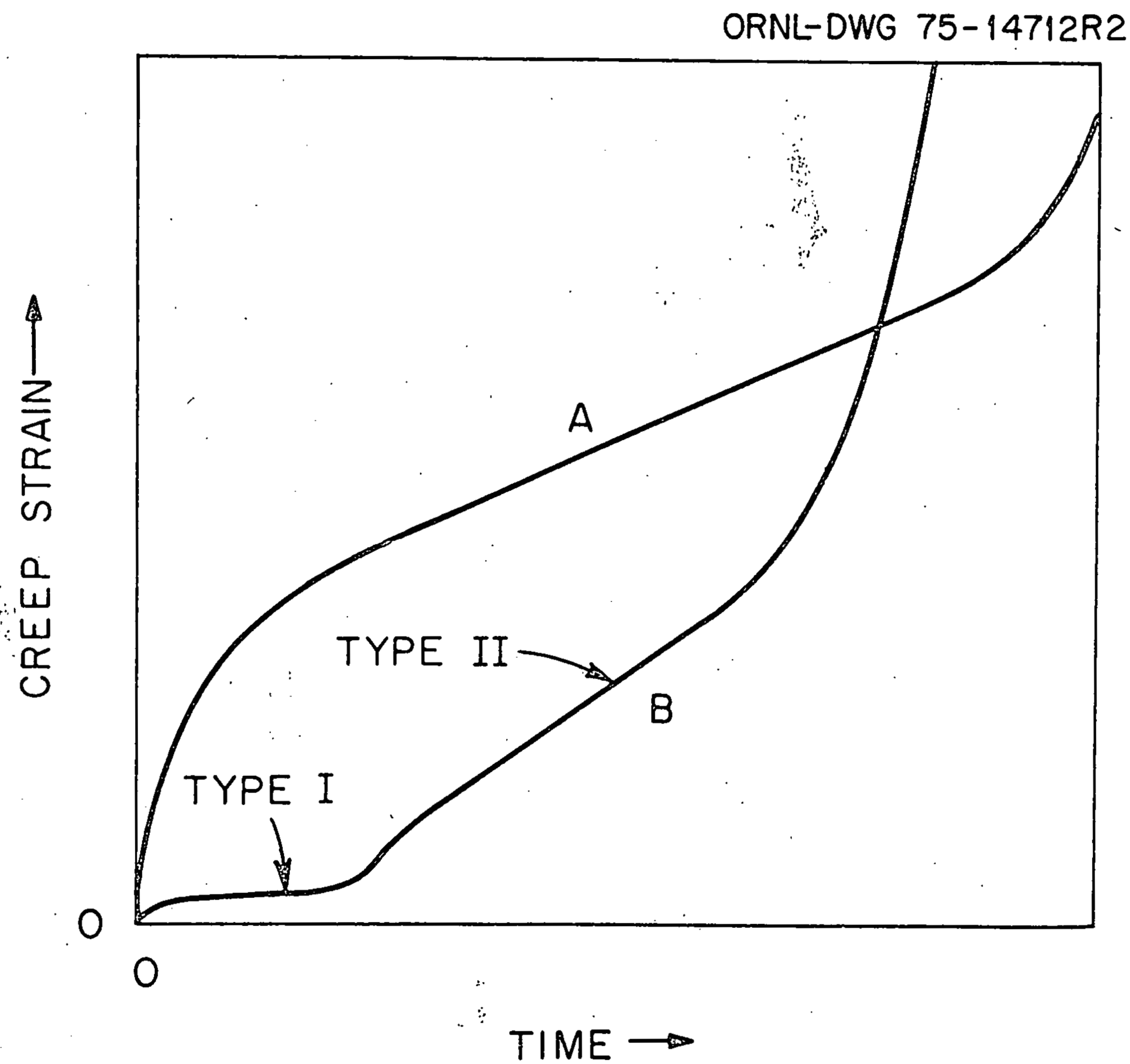
Table 5. Microhardness values from discontinued creep tests for alloy 718

Exposure time (hr)	Fraction of rupture life	Diamond[a] pyramid hardness
0	0.0	418
188	0.2	423
360	0.4	420
530	0.6	417
695	0.8	433
869	1.0	439

[a]Values represent the average of six measurements made within the specimen gage section. Due to microstructural inhomogeneities within the material, values typically ranged about 10 points above and below the stated average.

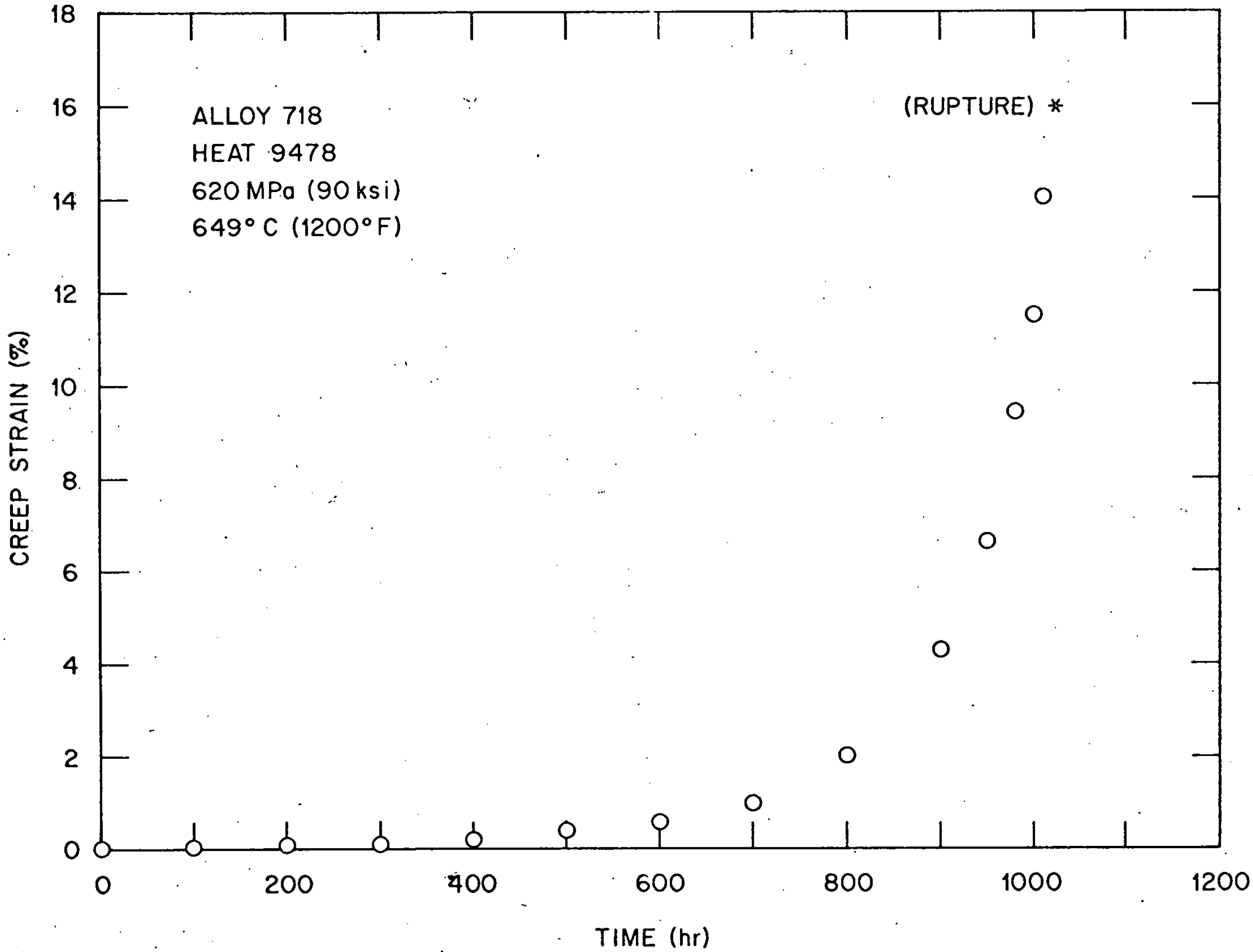


1. Schematic Definition of the Onset of Tertiary Creep for a Classical Creep Curve.

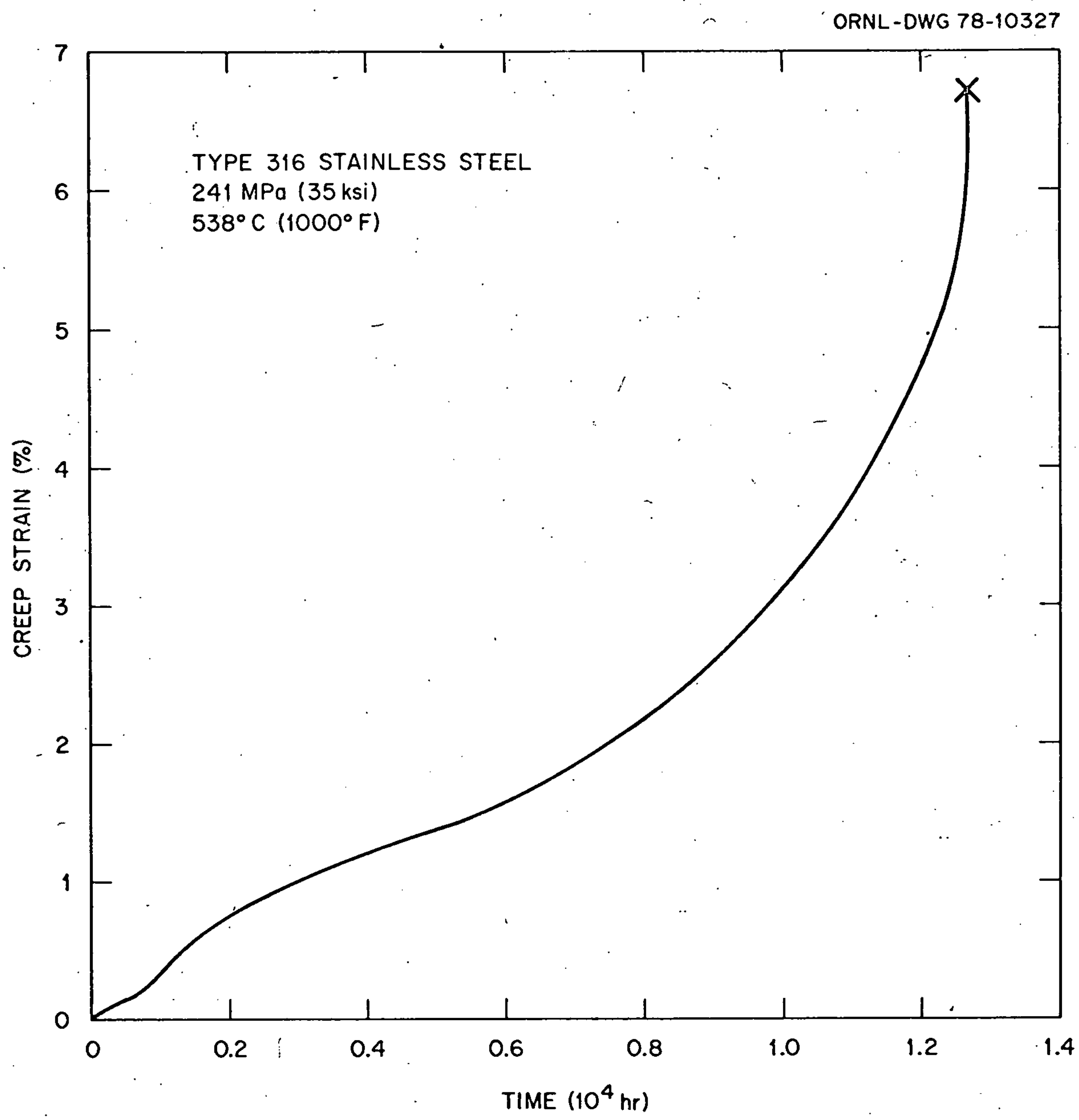


2. Illustration of Classical (A) and Nonclassical (B) Creep Behavior in 2 1/4 Cr—1 Mo Steel.

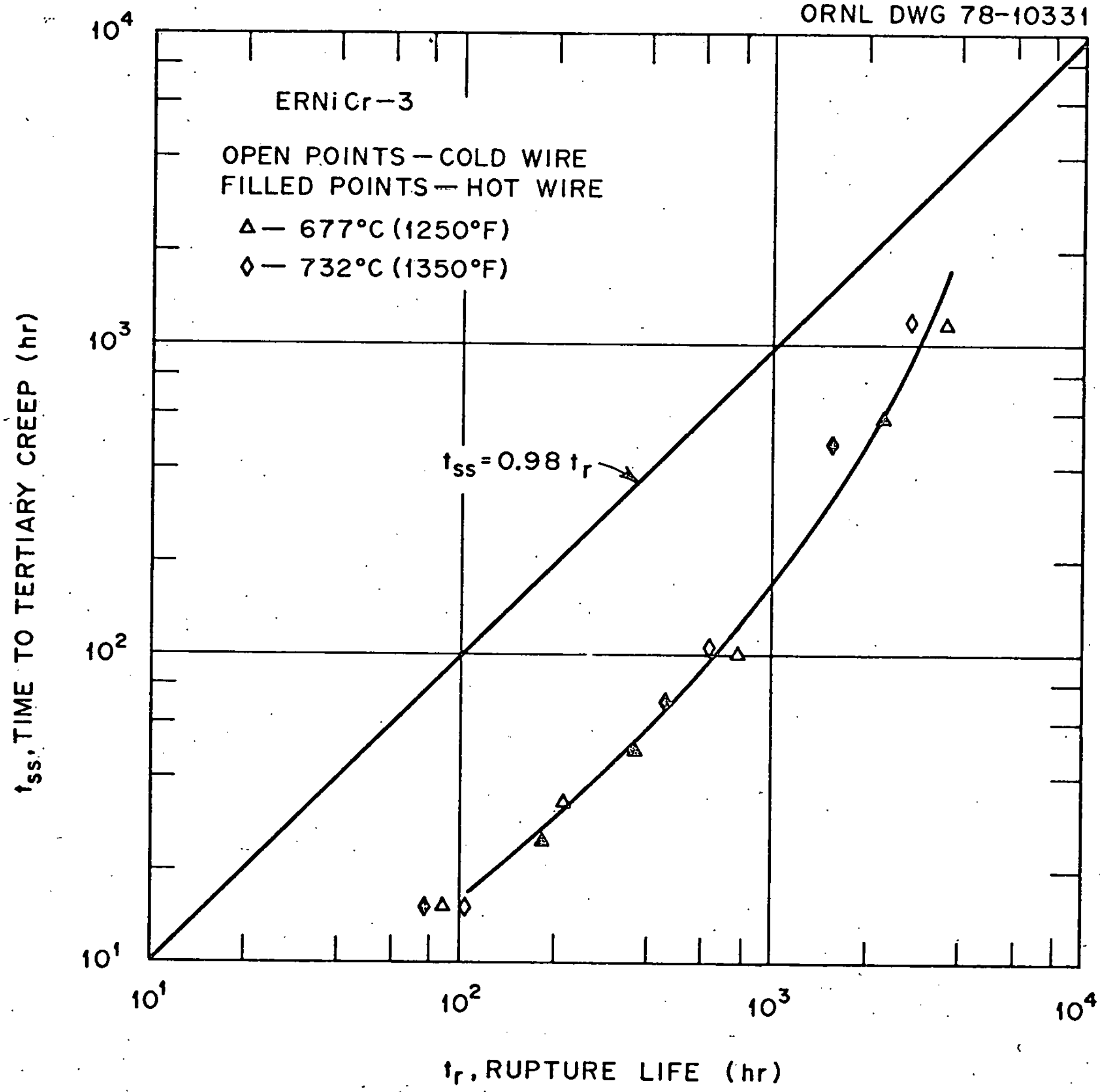
ORNL-DWG 78-10328



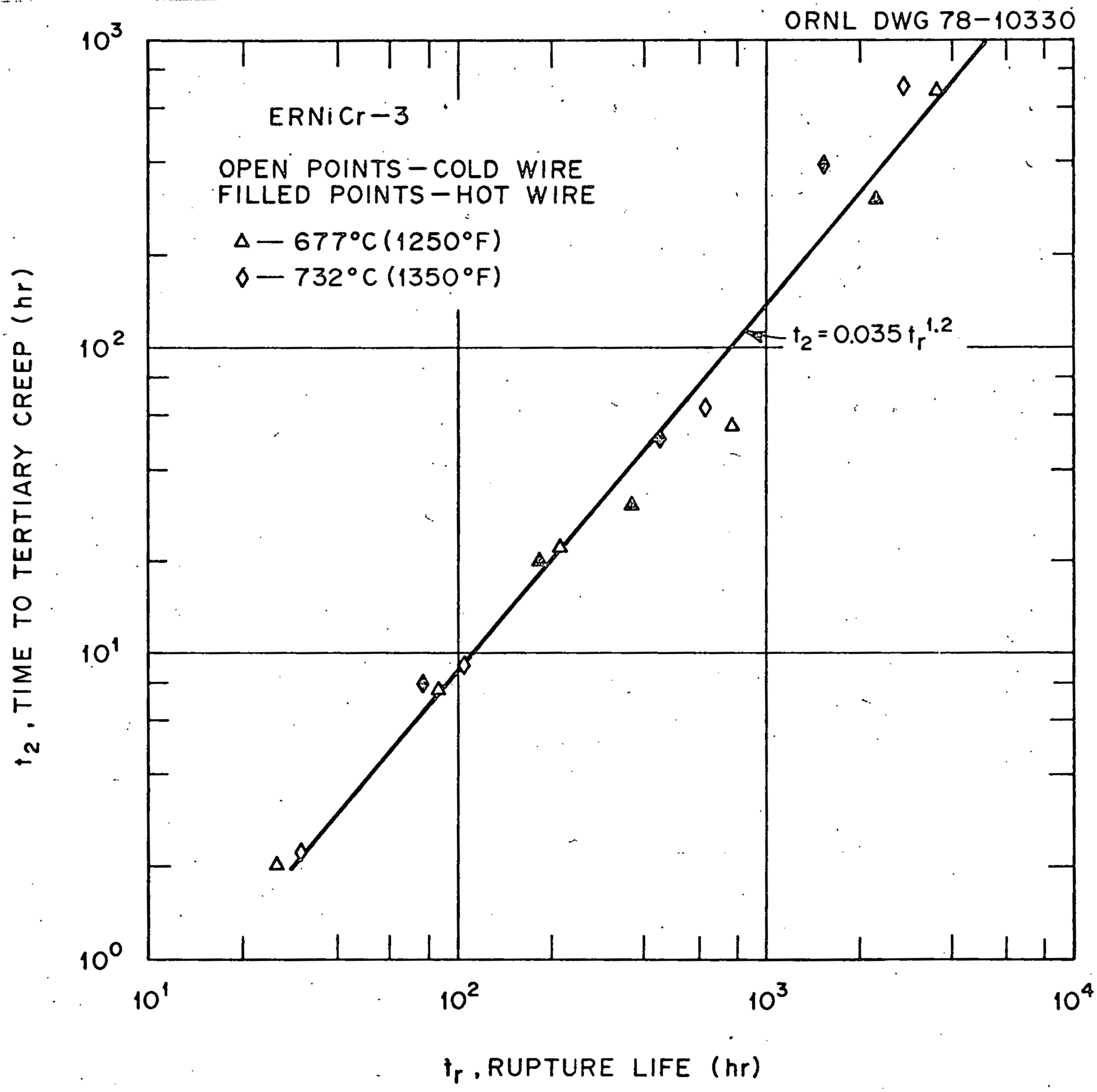
3. A Typical Creep Curve for Alloy 718.



4. Illustration of Nonclassical Creep Behavior in Type 316 Stainless Steel.

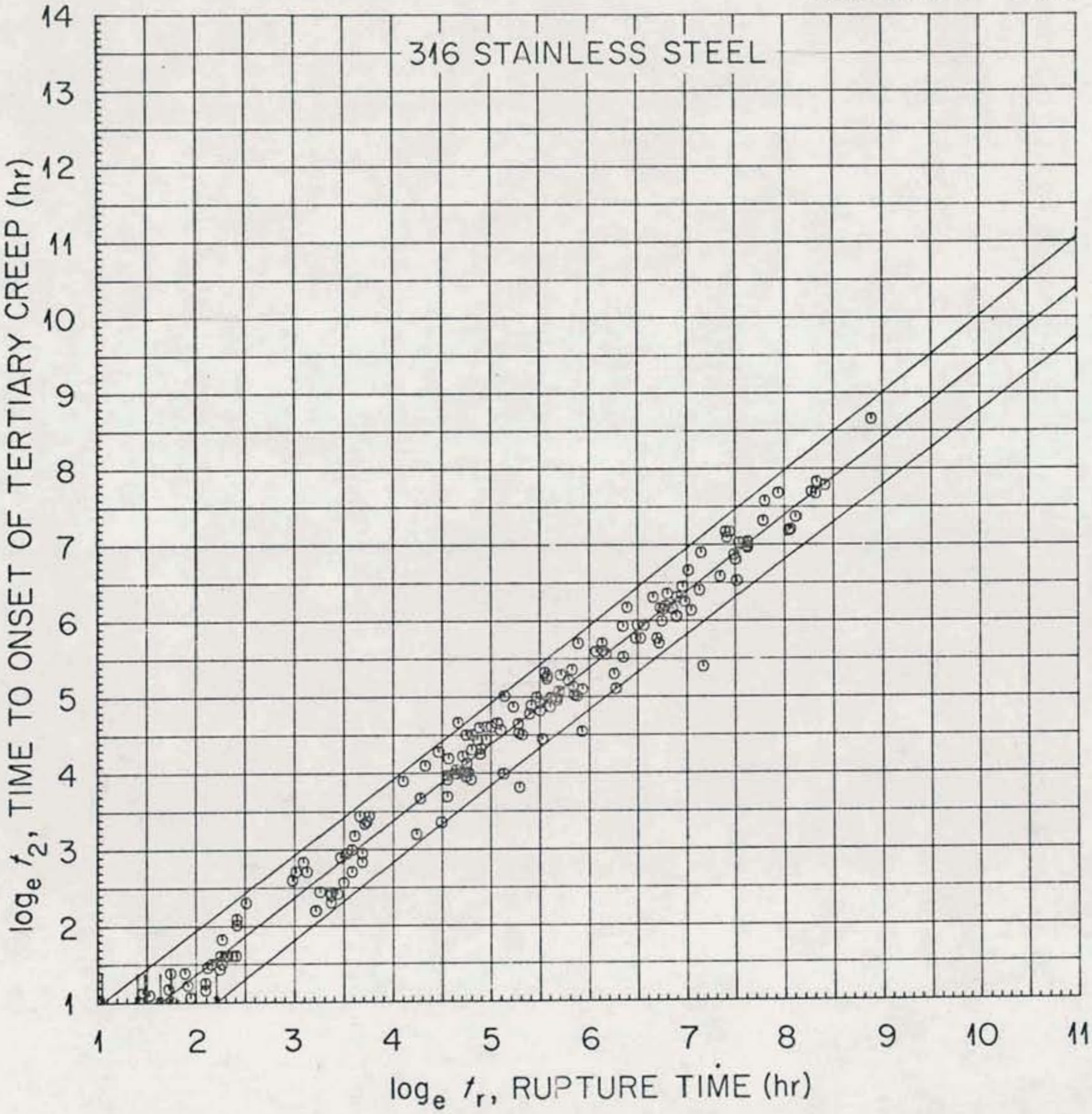


5. Relationship Between 0.2% Offset Time to Tertiary Creep and Rupture Life for ERNiCr-3 Weld Metal (677–732°C).



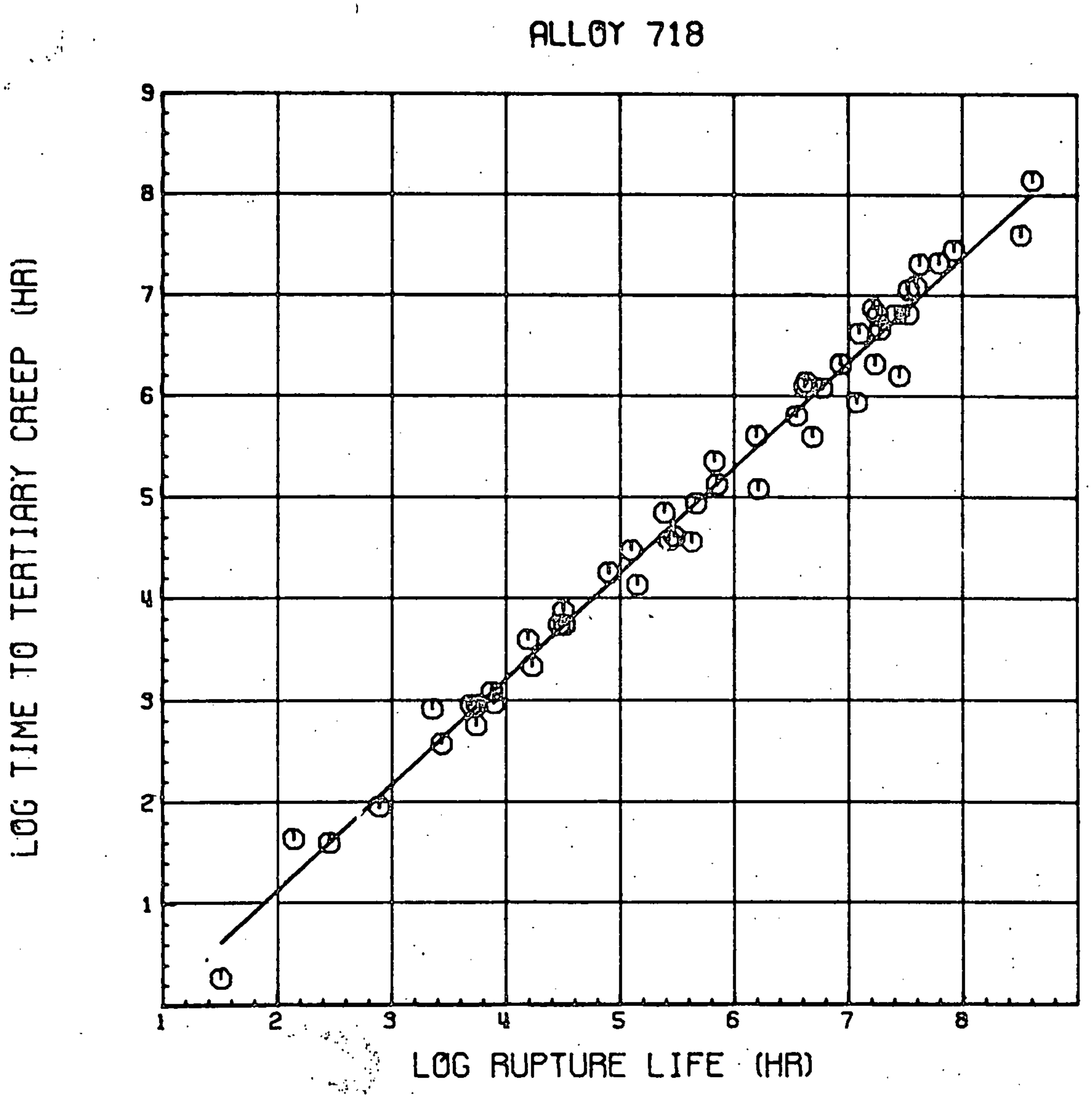
6. Relationship Between End of Linear Second Stage Creep and Rupture Life for ERNiCr-3 Weld Metal (677-732°C).

ORNL-DWG 75-2637



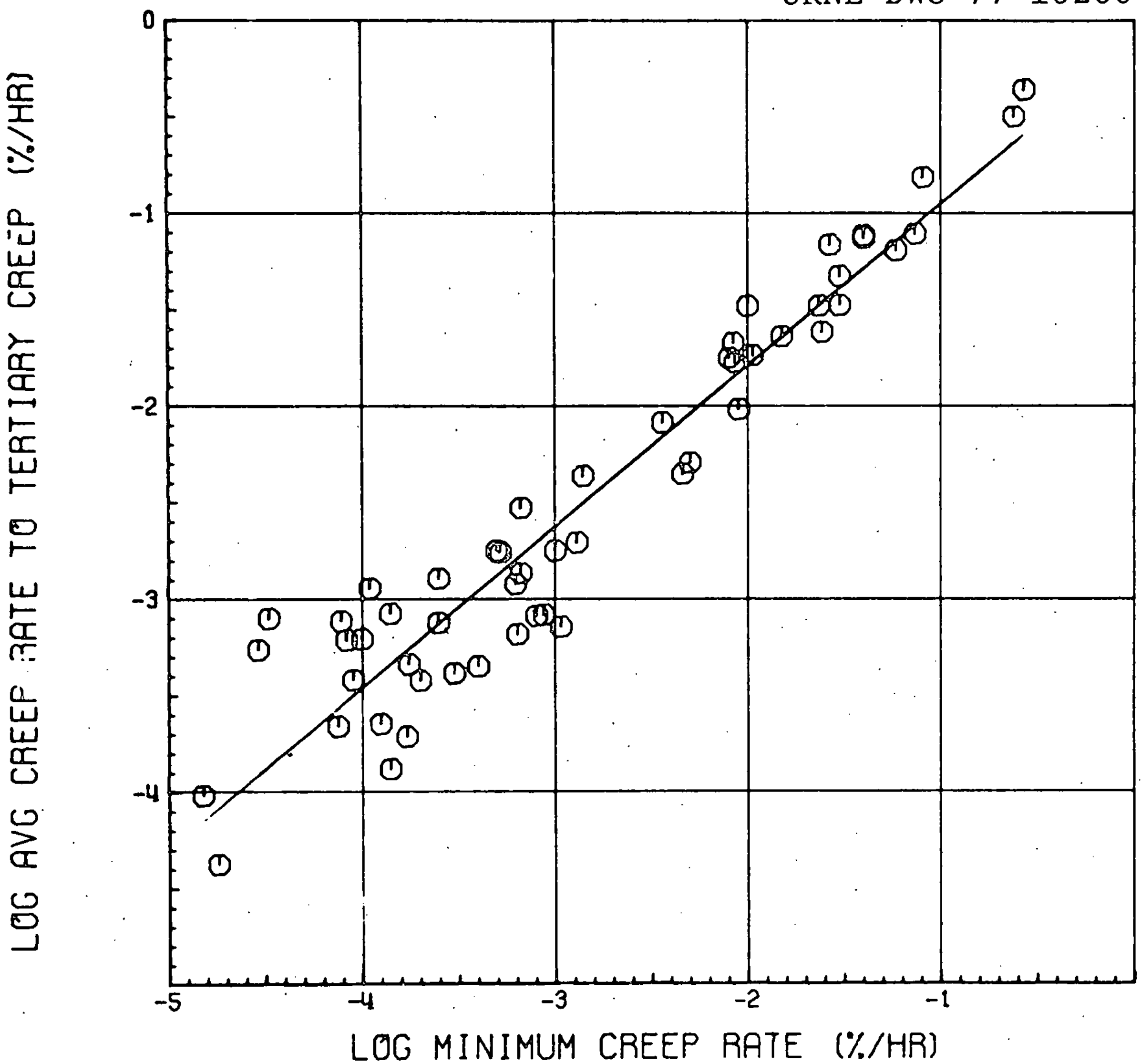
7. Relationship Between Time to Tertiary Creep and Rupture Life for Type 316 Stainless Steel.

ORNL-DWG 77-17069

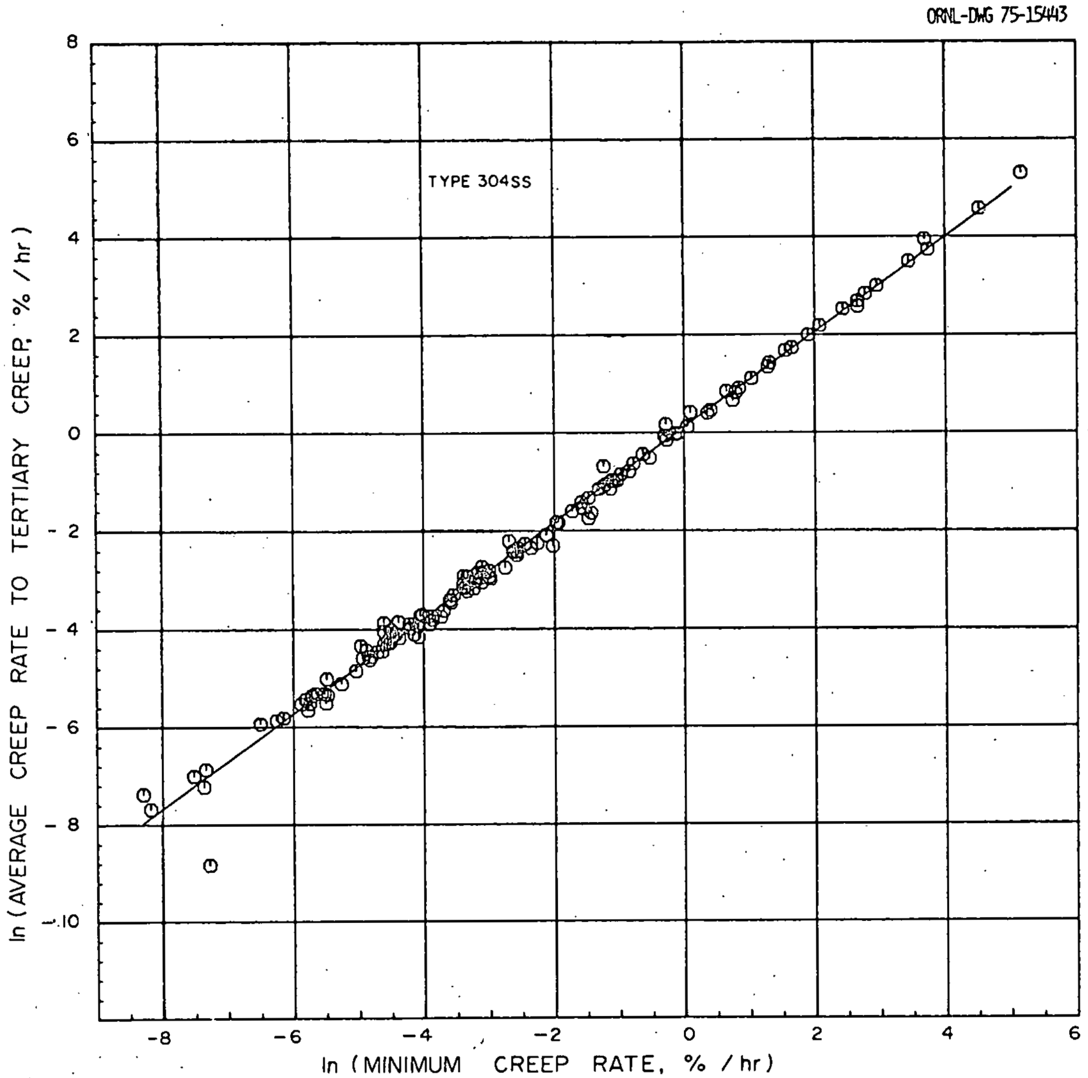


8. Relationship Between Time to Tertiary Creep and Rupture Life for Alloy 718.

ORNL-DWG 77-10266

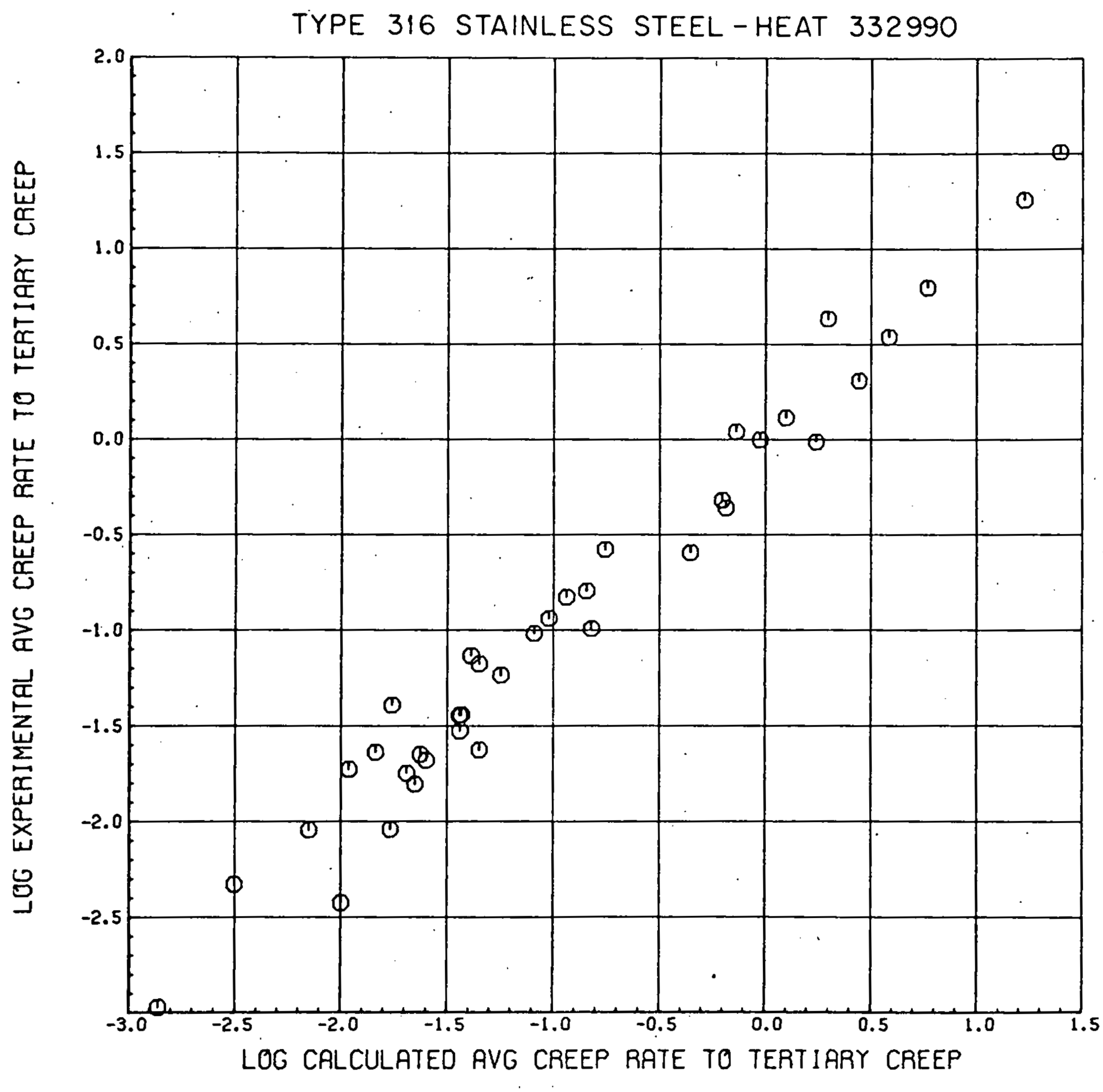


9. Relationship Between Average Creep Rate to Tertiary Creep and Minimum Creep Rate for Alloy 800H.

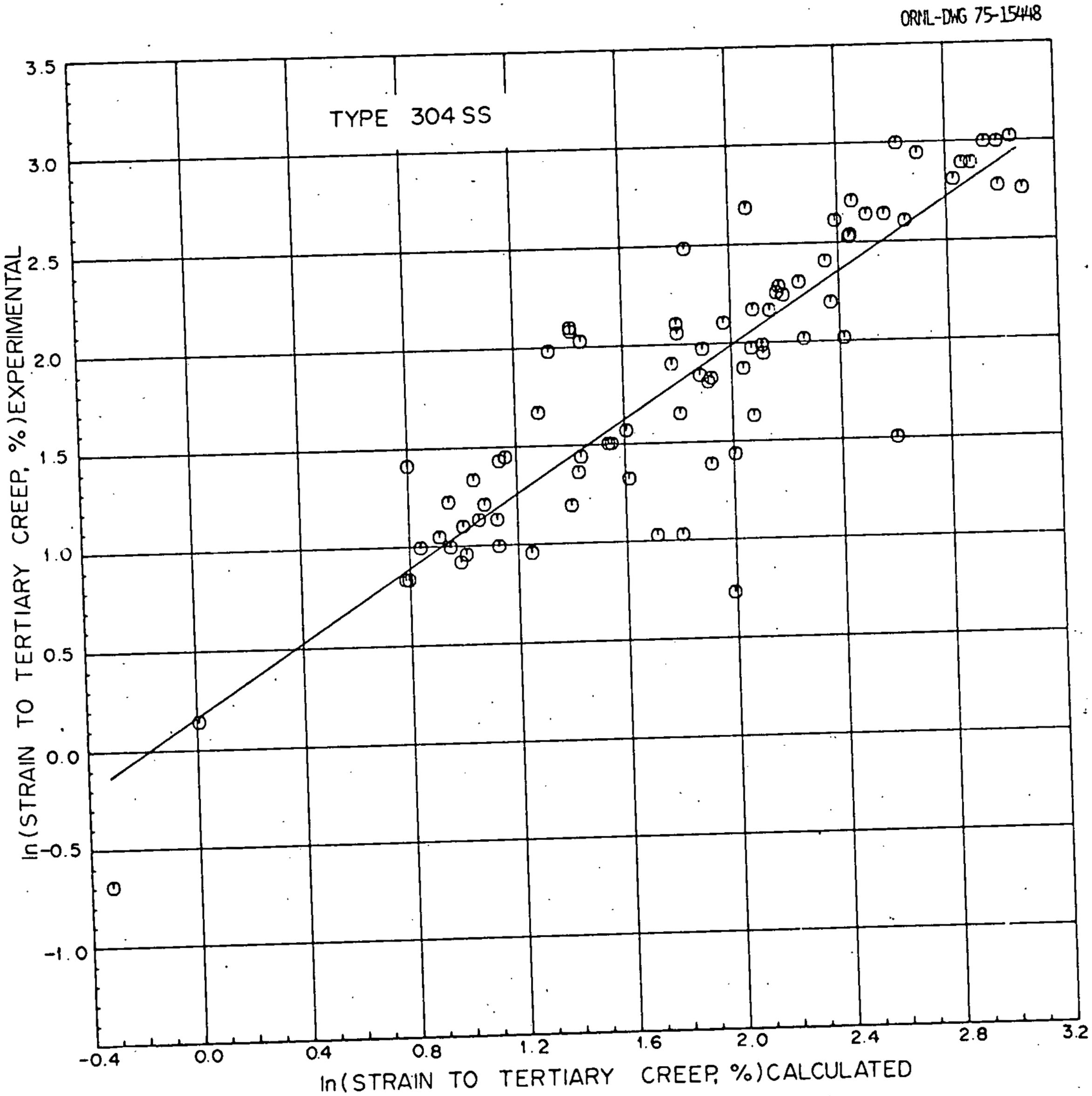


10. Relationship Between Average Creep Rate to Tertiary Creep and Minimum Creep Rate for Type 304 Stainless Steel.

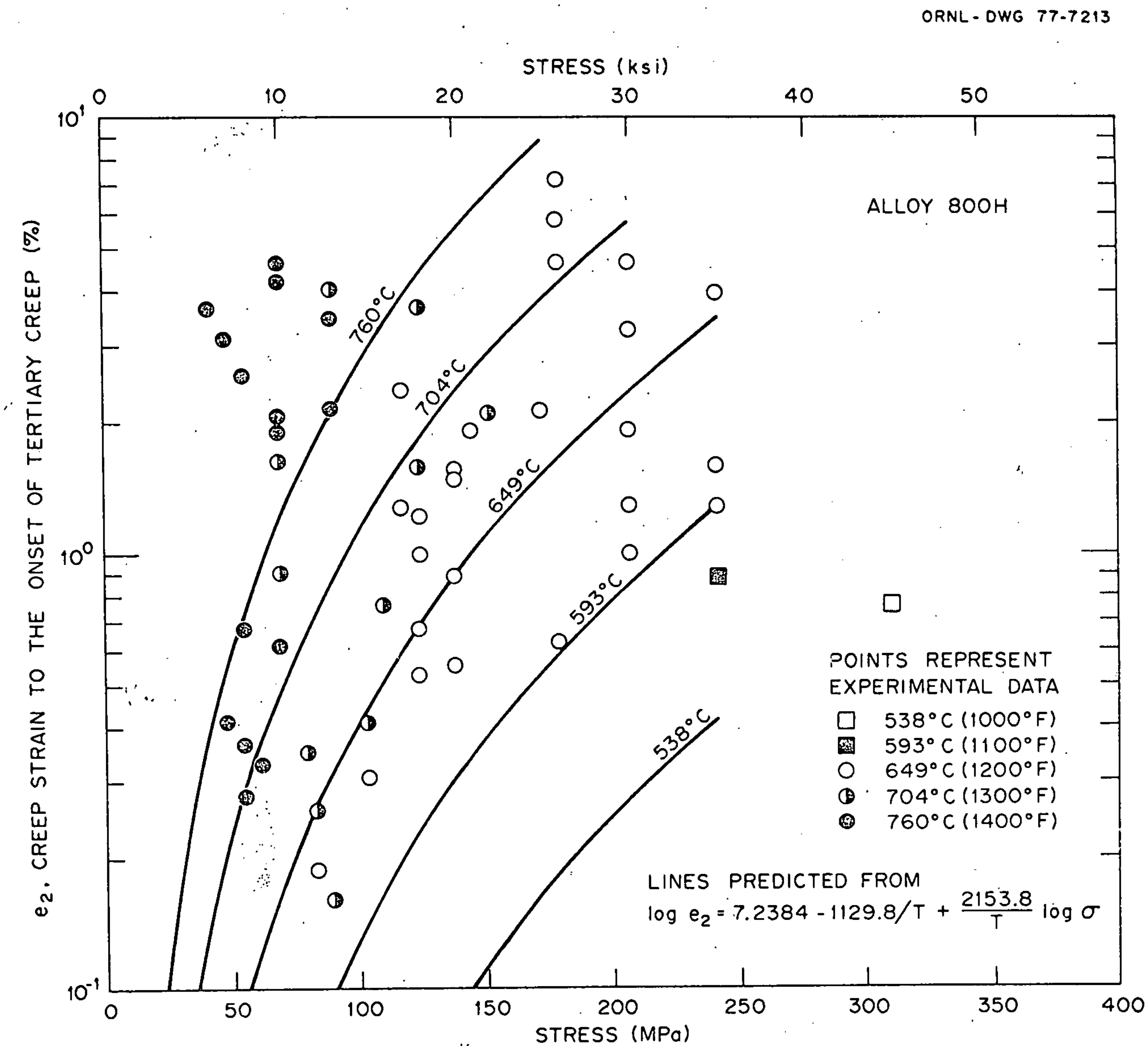
ORNL-DWG 76-4004



11. Comparison of Experimental Values for Average Creep Rate to Tertiary Creep with Values Predicted from $\dot{e}_3 = D_0 e^{-Q/RT} t_r^{-\gamma_0}$ for Type 316 Stainless Steel.

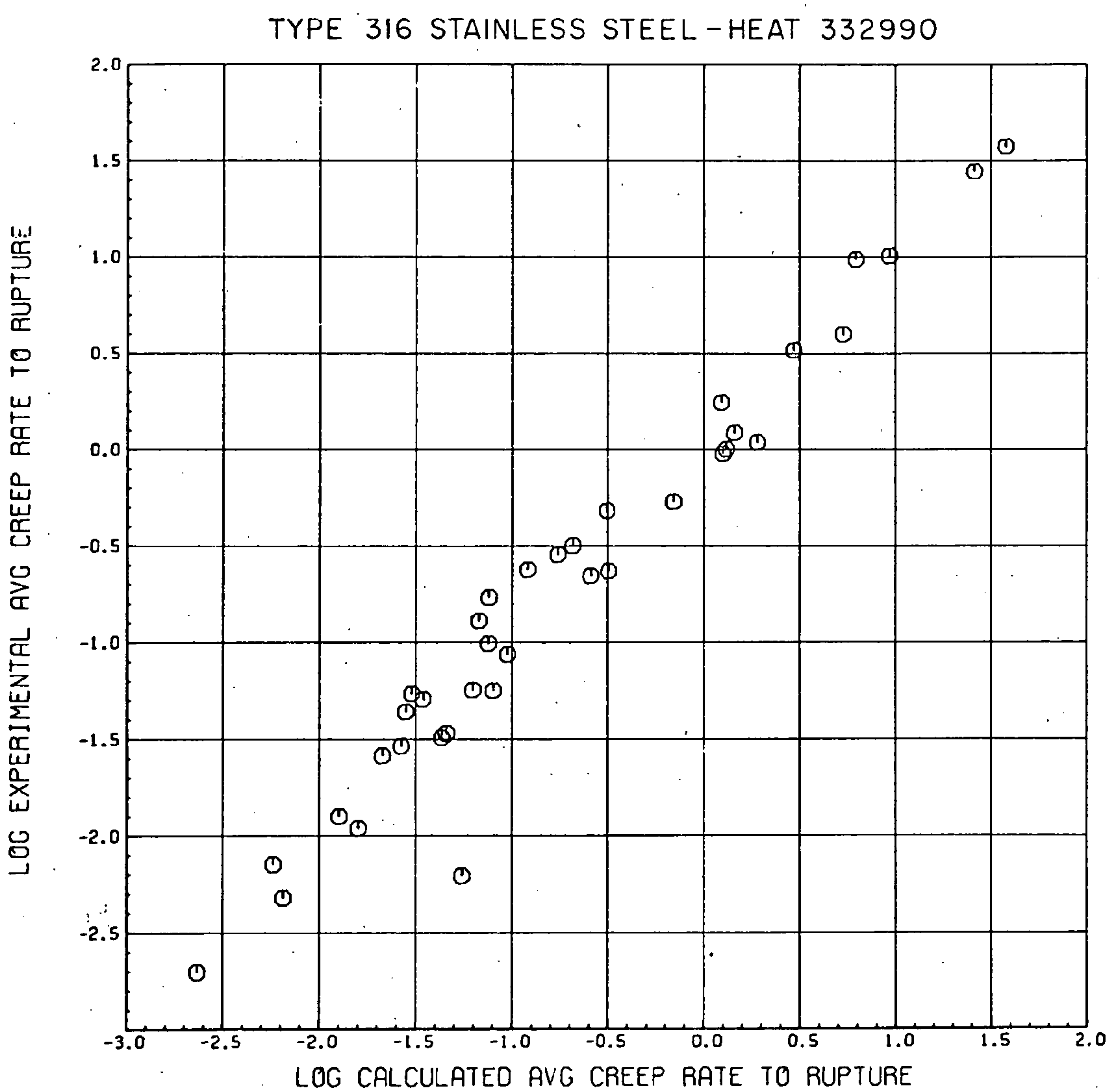


12. Comparison of Predicted and Experimental Creep Strain to Tertiary Creep for Type 304 Stainless Steel.

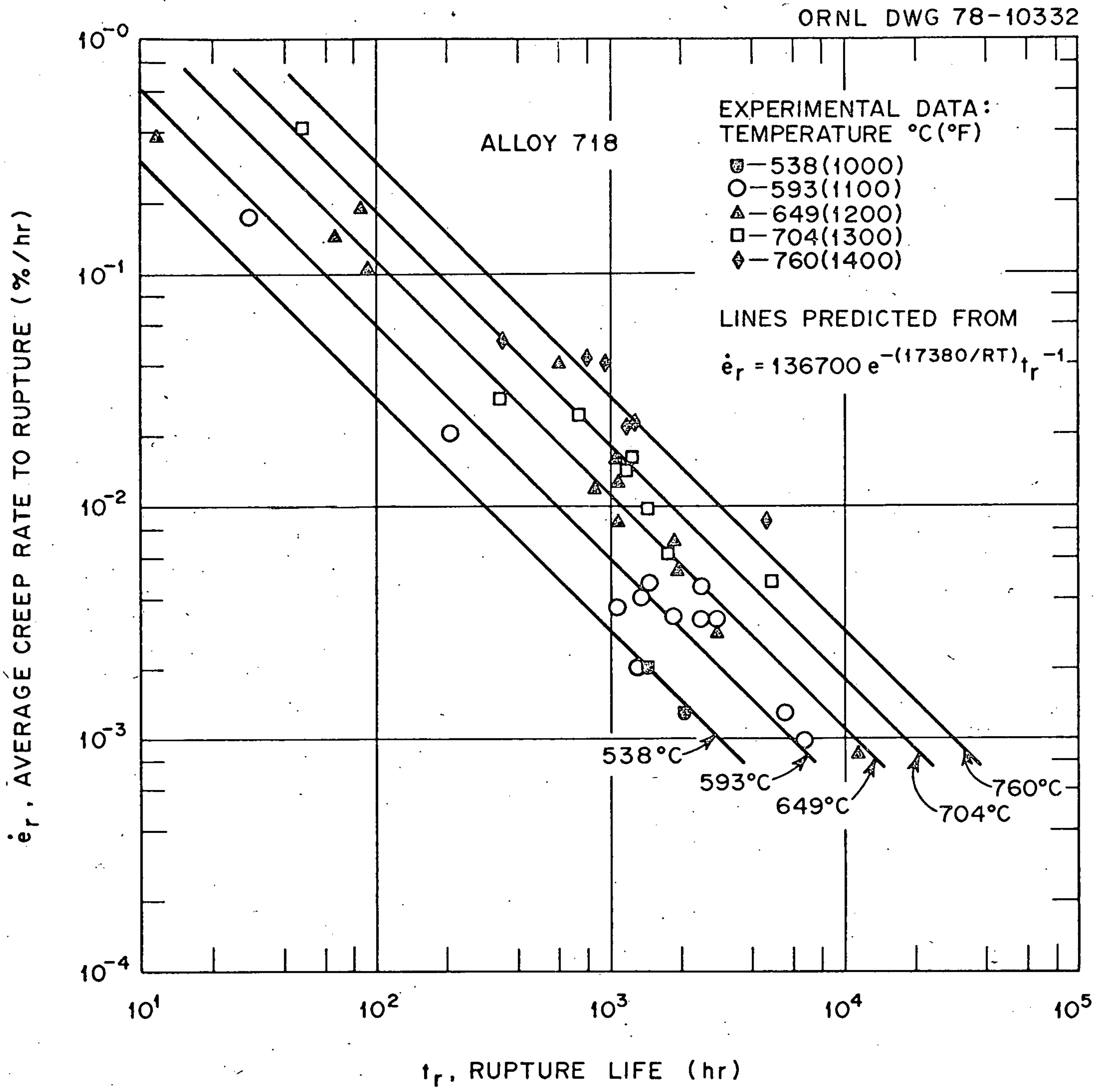


13. Comparison of Predicted and Experimental Creep Strain to Tertiary Creep for Alloy 800H.

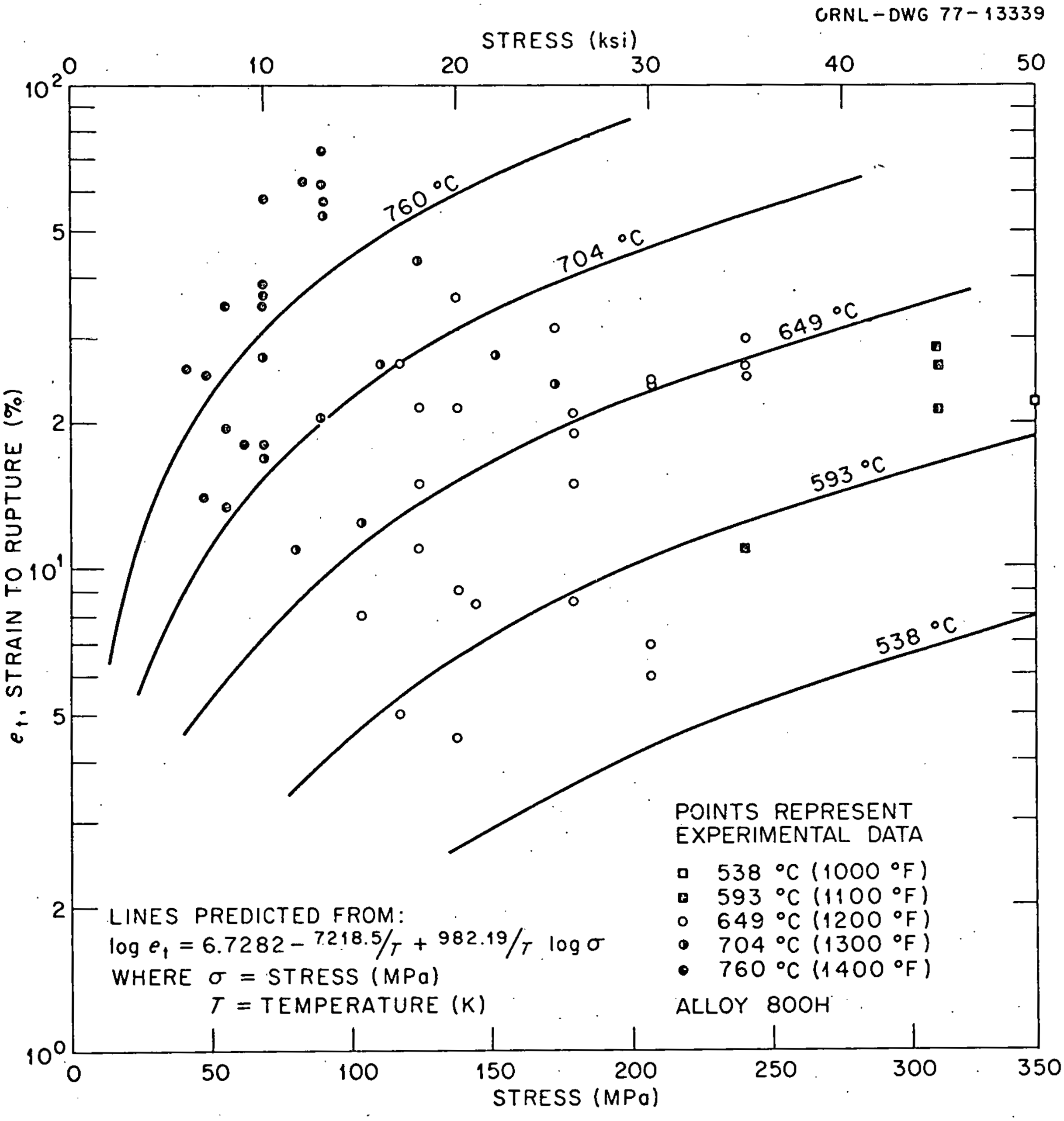
ORNL-DWG 76-4001



14. Comparison of Experimental Values for Average Creep Rate to Rupture with Values Predicted from $\dot{e}_r = E_0 e^{-Q/RT} t_r^{-\gamma_0}$ for Type 316 Stainless Steel.

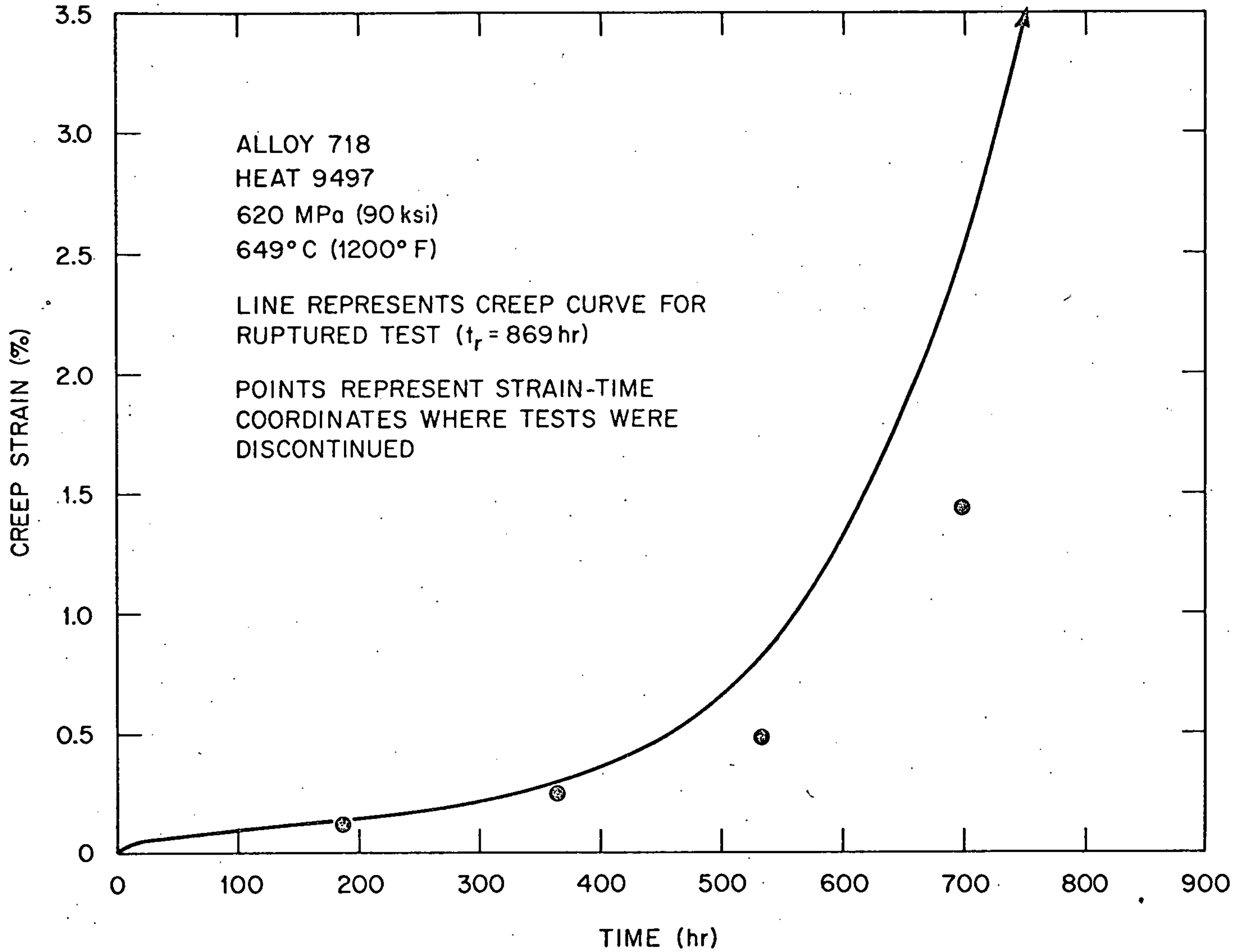


15. Relationship Between Average Creep Rate to Rupture and Rupture Life for Alloy 718.



16. Comparison of Predicted and Experimental Values for Strain to Rupture for Alloy 800H.

ORNL-DWG 78-10329



17. Creep Curves from Discontinued Creep Tests for Alloy 718.

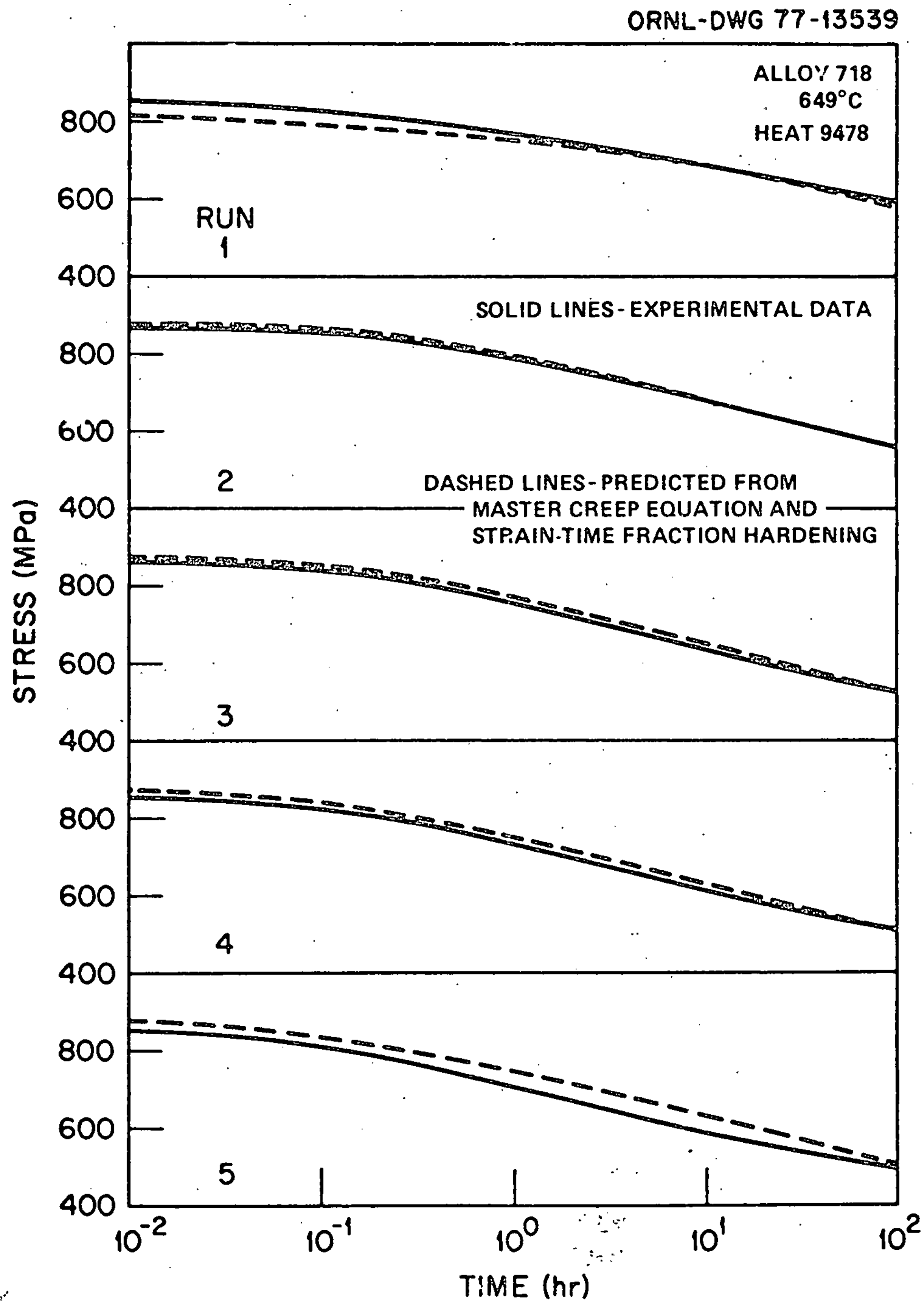
RUPTURED

DISC. 0.8 t_r

88411 500X

88410 500X

18. Typical Optical Microstructures from Ruptured and Discontinued Creep Tests for Alloy 718.



19. Comparison of Experimental Multiple Load Relaxation Data for Alloy 718 With Behavior Predicted from a Creep Equation and Strain Hardening.