CONF-850103--9

MASTER

LOAD-CARRYING CAPABILITIES OF REFRACTORY ALLOYS FOR SPACE REACTOR POWER APPLICATIONS*

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CONF-850103--9

DE85 010566

ABSTRACT

To achieve sufficient thermodynamic efficiency, space nuclear power systems must operate at temperatures above 1000°C. Such temperatures are too high for the successful use of commercially available alloys of iron, nickel, and cobalt. Only alloys in which refractory metals are the major components possess the required combination of long-term high-temperature strength properties required for fuel cladding and/or structural applications for space nuclear power systems. However, the experimental design data base for these alloys is very limited compared with the design data base available for iron-, nickel-, and cobalt-base alloys. Therefore, a quantitative evaluation of the existing mechanical properties data for the refractory alloys relevant to space nuclear power systems design lifetimes up to seven years at temperatures up to 1400°C is being conducted. The most important properties for space nuclear power systems are considered to be long-term high-temperature (>1000°C) creep strength and ductility, low-temperature (<400°C) fracture toughness [including ductile-to-brittle transition temperature, (DBTT)], and ductility at high strain rates; of special concern are the above properties for weldments of refractory alloys. Of principal interest at this time are creep properties as a function of alloy

*Research sponsored by the Office of Space Nuclear Projects, U.S. Department of Energy under contract DE-AC05-84OR21400 with the Martin Marietta Energy Systems, Inc.

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composition, applied stress, test temperature, test environment (e.g., vacuum, lithium), and thermomechanical treatment (TMT) history.

Currently being evaluated are, in order of ascending temperature of melting, selected alloys of niobium (e.g., Nb–1% Zr, Nb–1% Zr–0.1% C), molybdenum (e.g., Mo–13% Re), tantalum (e.g., ASTAR-811C), and tungsten (e.g. CVD W and W–25% Re). Creep properties of these alloys have been correlated via Larson-Miller, Manson-Haferd, and other empirical parameters; creep equations have been developed from these correlations. This evaluation assesses existing data and identifies the added data requirements necessary to assure success of the Space Nuclear Power Systems Program.

INTRODUCTION

Space nuclear power systems are designed to operate at temperatures in excess of 1000°C (~1275 K) to minimize the mass and volume of the power systems that must be launched. At temperatures above approximately 40% of the melting temperature (T_m), metals and alloys deform slowly by a process called "creep." For extended times at these temperatures (e.g., years) creep deformation, or creep strain, can become appreciable (e.g., 5–10%) at stresses that are a small fraction ($\lesssim$0.2) of the elevated temperature yield strength. For the proposed SP-100 power systems the reactor structures and fuel cladding are being designed to operate for 7 years (61,320 h) at temperatures between 0.4 and 0.5 T_m.

Table 1 lists the melting temperature and one-half the melting temperature for the base metals for several commercial structural alloys of nickel, cobalt, and iron, and for the refractory metals niobium, molybdenum, tantalum, and tungsten. Although the actual melting temperature of each alloy composition will vary from that of the pure metal, it will not vary sufficiently to change the order of melting shown in Table 1. Table 1 illustrates that only for alloys based on refractory metals is 1/2 T_m sufficiently high for long-term service at temperatures above 1000°C (~1275 K) to be considered feasible.

Information on the steady state load-carrying capabilities of the refractory alloys that are being considered for fuel cladding and reactor structural components of SP-100 can be established by determination of the creep strength and minimum creep rates at temperatures above 1000°C (~1275 K). Current conceptual designs indicate that a total creep strain on the order of 1 to 2% will limit the useful lifetime of fuel cladding and reactor structural components of the SP-100 power generation system (Hanson 1984); therefore, for this paper creep strength is assessed by the time required to produce 1 or 2% creep strain as a function of applied stress. Data on minimum creep rates are required to enable extrapolation of the strain produced in short-term creep tests to the strain produced in seven years since it is not feasible or practical, to conduct seven-year tests for all anticipated temperature-stress conditions. Until design criteria are established that provide quantitative strain limits for the components of the SP-100 reactor system, estimates of the stress required to produce 1% creep strain in seven years can be used to provide a ranking of the elevated temperature strength of candidate alloys. As stated earlier, only refractory metal alloys have sufficient strength at temperatures above 1000°C to limit deformation by creep to the required 1 to 2% in seven years. The purpose of this work is to characterize the high-temperature (>1000°C)

Table 1. Melting temperatures (T_m) and one-half the melting temperature (1/2 T_m) for elevated temperature structural metals

Metal	T_m (K)	1/2 T_m (K)
Nickel	1726	863
Cobalt	1768	884
Iron	1809	905
Niobium	2740	1370
Molybdenum	2890	1445
Tantalum	3287	1644
Tungsten	3680	1840

load-carrying capability of the refractory metal alloys considered as candidates for fuel cladding and/or structural applications for space nuclear power systems.

METHODS

To estimate the steady state load-carrying capabilities of refractory metal alloys for times relevant to SP-100 applications the existing creep data for several refractory metal alloys are being compiled, evaluated, and analyzed. The six materials currently under consideration are Nb-1% Zr, PWC-11 (Nb-1% Zr-0.09% C), Mo-13% Re, T-111 (Ta-8% W-2% Hf), ASTAR-811C (Ta-8% W-1% Re-0.7% Hf-0.025% C), W-25% Re, and chemically vapor deposited (CVD) tungsten.

Evaluation of the data is required to determine which data are to be analyzed for use in preparing the reference curve for creep strength versus temperature and the method of analysis. For example, refractory metals and their alloys are very reactive at high temperatures and have a high affinity for impurity atoms such as carbon, nitrogen, and oxygen. During creep testing in imperfect vacuua the carbon, nitrogen, and/or oxygen are absorbed into the alloy resulting in increased strength and decreased ductility of the alloy. Because lithium has a much higher affinity for these impurity atoms than do the refractory metal alloys and data from creep tests conducted in lithium are considered to be more representative of the actual creep strength of refractory metal alloys. This is especially valid for the more reactive niobium and tantalum base alloys. Therefore, the reference creep curve for these alloys is based on tests conducted in lithium, where possible. For Nb-1% Zr and PWC-11 the reference curves are based on tests in lithium. For T-111 and ASTAR-811C there are no data for tests in lithium and the reference curves for these two alloys are based on tests in vacuua of 10^{-9} to 10^{-8} torr.

Also, cold work in test samples affects the creep rate during testing at temperatures in excess of approximately 0.4 T_m. Therefore, for each alloy the reference creep curve is based on material that was fully recrystallized prior to testing.

The analysis is intended to provide an estimate of the maximum applied load that the above listed alloys can withstand and still result in acceptable creep strain under SP-100 operating temperature and lifetime requirements. Relative to the proposed SP-100 design lifetime of ten years with seven years of electrical power production most of the test data are for short-term creep tests. Therefore, parametric analyses such as Larson-Miller, Manson-Haferd, and Dorn methods are being used to predict from the relatively short-term data, the applied stresses that will produce 1% (or similar) creep strain in seven years.

Analyses are also in progress to attempt to estimate the load-carrying capability of these seven materials under conditions of cyclic loading with long hold times (10^2–10^4 h) at high temperature between cycles (creep-fatigue). For alloys for which there is data such as types 304 (Maiya 1981) and 316 (Hales 1980) stainless steel, 2 1/4 Cr-1 Mo steel (Brinkman 1983) and alloy 718 (Thakker and Cowles 1983) periodic cycling of the creep load (creep-fatigue) significantly decreases the load-carrying capability relative to that for monotonic creep loading. No creep-fatigue data are available for refractory metal alloys; however, since the SP-100 is being designed to produce electrical power for a cumulative time of seven years during a ten year mission the load-carrying capabilities of these alloys may be appreciably lower than predicted from monotonic creep data. For the space power systems that are based on the Stirling engine concept, the load carrying capabilities under high-cycle fatigue (>10^8 cycles) loading must be evaluated. For at least one of the proposed space reactor power systems the Stirling engines are to be operated at a frequency of approximately 100 Hz. For seven years of operation these engines and their components will be subjected to approximately 2.2×10^{10} cycles of low strain range fatigue loading. Since the stress at which failure can occur due to high-cycle fatigue, loading may be below the stress to produce 1% creep strain in seven years the high-cycle fatigue properties of these seven materials should be evaluated.

Also, data on the high-temperature (>1000°C) creep properties and low temperature (<250°C) ductility and toughness properties of weldments of these seven materials are being compiled and evaluated.

RESULTS AND DISCUSSION

At this time analyses using the Larson-Miller parameter have been completed for Nb-1% Zr, PWC-11, T-111, and ASTAR-811C. Analyses are in progress for these alloys using the Manson-Haferd and Dorn parameters. Data are being compiled for Mo-13% Re, CVD-W, and W-25% Re.

Nb-1% Zr

Table 2 contains pertinent information on the data obtained for Nb-1% Zr. As shown in Table 2, not all of the data compiled have been used

Table 2. Available data for creep tests on Nb-1% Zr

Reference	No. of tests	Tests used for reference data base	Total test time (h)	Total time of tests used (h)	Longest test (h)	Longest test time used (h)
PWAC-1963	14[a]	13	131,839[b]	129,606	10,037[c]	10,037
PWAC-1964a	31	0[d]	21,299	0	2,511[e]	0
McCoy-1964	21[f]	0[d]	9,310	0	1,733[e]	0
Chang-1962	11[g]	0[d]	476	0	109[h] (10^{-6} torr)	0
Stewart et al., 1961	27[i]	0[d]	9,532	5,162	1,196[e]	1,196
Hall and Titran, 1960	2	0[d]	307	0	300[h]	0
TOTALS	105	71	172,763	134,768		

[a]Includes three welds.

[b]Time to rupture or discontinuance of test.

[c]All tests of ~10^4 h in lithium.

[d]Tests plotted with reference line in Fig. 1.

[e]Tested in lithium.

[f]Thirteen samples tested with >40% cold work; eight other samples tested after various annealing treatments.

[g]Five samples cold worked; six others with various annealing treatments.

[h]Tested in vacuum.

[i]Eleven samples cold worked.

in the analysis. The number of tests, total test time, and longest single test are used to assess the relevance and reliability and to establish a relative figure of merit for extrapolating the data to SP-100 conditions. Figure 1 is a plot of applied stress to produce 1% creep strain as a function of Larson-Miller parameter for Nb-1% Zr. Minimum creep rate is in percent per hour and the temperature scales on the top abscissa are for 61,320 h (7 years). The reference line is based on 13 tests in lithium, each of about 10^4 h in duration (PWAC-1963). The tests on annealed material are shown by the open circles. For all figures in this paper the filled symbols represent the average Larson-Miller parameter for multiple tests at the same stress.

The reference line includes two tests of weldments (open diamonds) in lithium which were also of about 10^4 h duration (PWAC-1963). The weldment tests are included because the tests were identical to those for base metal (10^4 h in Li), the data for them is so close to that for the base metal that there is no obvious reason for excluding them, and including them increases the data base by 18%. The third data point for a weld tested at a stress of 13.8 MPa (2 ksi) and Larson-Miller parameter of 24.7×10^3 is not included in the reference line because the sample failed after only 2233 h with 44.3% elongation (PWAC-1963). Therefore, it is not included with the tests that exhibited approximately 1 to 5% strain in 10^4 h. The data points for Fig. 1 were obtained by dividing the total measured strain by the recorded time to give an "average steady state" or minimum creep rate. In the absence of measured times to 1% creep strain or measured minimum creep rate this appears to be the best method of using the data available for calculating

Larson-Miller parameters. Since all of the tests used to establish the reference line in Fig. 1 were for about 10^4 h duration, increased values of Larson-Miller parameter are due only to increased test temperatures.

The limited data in Fig. 1 indicate that for design applications, welded structures of Nb-1% Zr can be considered to have the same elevated temperature creep properties as base metal.

In addition to the above tests an additional 16 tests in lithium on annealed Nb-1% Zr are shown in Fig. 1 (Stewart et. al 1961). These tests were for much shorter times (14—1196 h) than were the above listed tests and more than half of the tests were conducted at 982°C (1255 K). As a result of the shorter times and relatively low temperature of the tests, the values for the Larson-Miller parameters are grouped at the lower values from 18×10^3 to 23×10^3. Over this range the data are equally distributed above and below the extrapolation of the reference line.

Figure 2 shows additional data for Nb-1% Zr tested in lithium (PWAC-1964a). These data are for tests that were of much shorter time (6—2511 h) than those shown in Fig. 1. However, for test temperatures of 871°C (1144 K) and above, the shorter term data are in good agreement with the reference line. There were no data for the reference line below 871°C (1144 K). Note as in Fig. 1 that the weld specimens have the same creep strength as the base metal.

Figure 3 shows data for Nb-1% Zr tested in vacuua (Chang 1962, McCoy 1964, Hall and Titran 1966). Much of the data shown in Fig. 3 are for material in the cold worked condition (>40%). As expected, annealing of the cold worked material increases the creep strength and the increase in creep strength increases with increased annealing temperature for the range of annealing temperatures reported by McCoy (1964). For data reported by Chang (1962) there is little change in creep strength between the cold worked and annealed conditions. This indicates that the level of cold work in these samples probably was rather small compared to the level of cold work for the material described by McCoy (1964). All of the data reported by Chang (1962) are above the reference line. This is most probably due to impurity atom contamination and concomitant strengthening due to testing in imperfect vacuua.

PWC-11 (Nb-1% Zr-0.09% C)

Table 3 contains pertinent information on the data obtained for PWC-11. The initial PWC-11 creep data evaluated were based on 35 tests conducted in lithium (PWAC-1964b, 1965a). Twenty-eight tests were performed on material with the same thermomechanical treatment (TMT) history. All samples were extruded at a temperature between 1649 and 1816°C (1920 and 2090 K) and subsequently annealed for 2 h at 1204°C (1477 K). The remaining seven tests were performed on material with the three additional TMTs shown in Fig. 4. These data were evaluated using the Larson-Miller parameter and reported minimum creep rates. A curve visually fitted through these data is provided in Fig. 4.

The material utilized for these 35 tests had three different carbon contents, 550, 800, and 900 ppm by weight (PWAC-1965b). Data for all three carbon levels are included in the analysis to establish the reference curve.

Table 3. Available data for creep tests on PWC-11

Reference	No. of tests	Tests used for reference data base	Total test time (h)	Total time of tests used (h)	Longest test (h)	Longest test time used (h)
PWAC-1964b, 1965b, and 1965c	35[a]	28	39,878[b]	26,343	3,861[c]	3,861
PWAC-1965c	37[d]	0	1,880	0	357[e]	0
TOTALS	72	28	41,758	26,343		

[a]Twenty-eight tests on material with same thermomechanical treatment (TMT) history, extruded at 1649 to 1816°C and annealed for 1 h at 1204°C; seven tests on material with seven different TMTs.

[b]Time to rupture or to discontinuance of test.

[c]Tested in lithium.

[d]The thirty-seven tests were on material with 28 different TMTs and time to 1% creep strain not reported.

[e]Tested in unspecified vacuua.

As shown in Fig. 5, carbon content (over the above range) had no effect on the shape of the curve and only a small effect on the magnitude of the Larson-Miller parameter versus applied stress as shown by the test temperature information provided at the top of Fig. 5. The shape of the stress to produce 1% creep strain curve is determined by test temperature.

More recent information relating the time to 1% creep strain for 18 of the 35 tests shown in Fig. 4 was located (Watson 1964). The time in hours for 1% creep strain as a function of applied stress for the 18 tests has been evaluated using the Larson-Miller parameter. A curve visually fitted through these data is provided in Fig. 6. Because this curve is based on measured time to 1% creep strain it does include strain contributions associated with primary creep and is considered a more conservative estimate of the creep response of this alloy. The curve based on the minimum creep rate data from Fig. 4 for these 18 tests has been included in Fig. 6. At lower values of Larson-Miller parameters ($\lesssim 21 \times 10^3$) the curves based on minimum creep rate data and for time to 1% creep strain are in reasonable agreement. At higher values ($\gtrsim 23 \times 10^3$) the curve based on minimum creep rate data is slightly above the curve based on the time to 1% creep strain.

Both curves in Fig. 6 becomes nonlinear at the higher values ($\gtrsim 23 \times 10^3$) of Larson-Miller parameter which is generally associated with higher creep testing temperatures. This curvature at the higher temperatures suggests that thermal aging (carbide phase particle growth) could be occurring during the creep test. The curve is essentially linear at lower temperatures where thermal aging should not be occurring during the times associated with these tests.

Data on yield strength and tensile ductility after exposure at temperatures of 1060°C (1333 K), 1010°C (1283 K), and 730°C (1003 K) for 1000 and 3000 h shown in Table 4 provide an additional indication of the thermal instability of the alloy (PWAC, 1965a). After thermal exposure for 1000 h in argon and 3000 h in vacuum, the yield strength at 1060°C (1335 K) has been decreased by approximately 50% relative to that of the unexposed material and the ductility has been increased by the same amount. At a test temperature of 1010°C (1285 K) there are no data for unexposed material in PWAC 1965a; however, increasing the exposure time from 1000 to 3000 h decreased the yield strength by 50% and increased the ductility by the same amount. Aging for 3000 h in vacuum at 730°C (1005 K) resulted in little or no increase in yield strength but a 33% decrease in elongation (Table 4).

Data on the minimum creep rate as a function of applied stress for a total of 72 tests conducted either in lithium (PWAC 1964b, Watson 1964) or in vacuum (PWAC 1965c) have been analyzed. The 35 PWC-11 creep tests in lithium were conducted at temperatures from 871°C (1144 K) to 1315°C (1589 K) and applied stresses from 12.4 MPa (1.8 ksi) to 207 MPa (30 ksi). A plot of minimum creep rate ($\dot{\varepsilon}_{min}$) as a function of applied stress (σ_a) for these 35 tests (Fig. 7) suggests a change in creep response of the PWC-11 alloy at approximately 1093°C (1366 K). The slopes of $\dot{\varepsilon}_{min}$ for tests conducted below 1093°C (1366 K) are significantly less than for tests conducted at 1093°C (1366 K) and above. Also, for the tests at 1093°C (1366 K) and

Table 4. Effects of thermal exposure on tensile properties of PWC-11[a]

Specimen history	Temperature[b]					
	1060°C (1335 K)		1010°C (1285 K)		730°C (1005 K)	
	Yield strength (ksi)	% Elongation	Yield strength (ksi)	% Elongation	Yield strength (ksi)	% Elongation
10–20% cold work followed by 1320°C, 1 h and 1250°C, 1 h	23.4	19.5				
1000 h in argon	12.9	29	21.7	20		
3000 h in vacuum	10.0	33	11.7	29.5	24.2	13

[a]Data from Table 1, p. 59, PWAC-1965a, January 27, 1965.

[b]Temperatures are the exposure temperatures and subsequent tensile test temperatures.

above the slope increases with increasing test temperature. The differences in slopes of $\dot{\varepsilon}_{min}$ for the two temperature ranges may be due to thermal aging that occurs in this alloy at elevated temperatures.

For the creep tests at 871 and 982°C (1144 and 1255 K) aging probably did not occur during the time duration of the creep tests because maximum duration of the tests at 871 and 982°C (1144 and 1255 K) was 2702 and 617 h, respectively. Based on the changes in yield strength and ductility for exposures at 1010 and 1060°C (1285 and 1335 K) (see Table 4), aging probably occurred during the creep tests at 1093°C (1366 K) and above. The increased slopes of the $\dot{\varepsilon}_{min}$ as a function of applied stress with increasing temperature is a further indication of aging in the alloy.

Also, it has been reported that aging this alloy in static argon or flowing lithium prior to creep-rupture testing reduced the creep-rupture life (PWAC-1964b). Creep-rupture tests were conducted at a temperature of 1204°C (1477 K) and a stress of 27.5 MPa (4 ksi) on material that had been solution-annealed for 1 h at 1579°C (1852 K) or at 1593°C (1866 K) and annealed for 2 h at 1204°C (1977 K) and on material that received the same heat treatment and then exposed to static argon or flowing lithium for 1000 h at 1204°C (1477 K) prior to creep rupture testing. As shown in Table 5 for the tubing with the 0.025 in. wall thickness, the 1000 h pre-exposure reduced the creep-rupture life from 875 to 386 h for material aged in static argon and to 263 h for material aged in flowing lithium (PWAC 1964b). For the tubing with the 0.013 in. wall thickness, the 1000 h pre-exposure reduced the

Table 5. Effect of aging on creep-rupture life of PWC-11 tubing[a]

Aging time at 1204°C	Creep-rupture life (h)[b]	
	Test 1[c]	Test 2[d]
None	935	875
10^3 h in static argon	690	386
10^3 h in flowing lithium	78[e]	263

[a]Data from PWAC-643, Table 4, pp. 179 and 181, November 4, 1984.

[b]Tested at 1200°C and 27.5 MPa (4 ksi).

[c]Tubing 0.312 in. OD × 0.013 in. wall, annealed 1 h at 1593°C + 2 h at 1204°C.

[d]Tubing 0.250 in. OD × 0.025 in. wall, annealed 1 h at 1579°C + 2 h at 1204°C.

[e]Specimens may not have received annealing treatment.

creep rupture life from 935 to 690 h for the material aged in static argon and to only 78 h for material aged in flowing lithium (PWAC 1964b).

Note that the thinner wall thickness (0.013 in.) tubing suffered a much larger decrease in creep-rupture life than did the thicker wall thickness (0.025 in.) tubing. Since the two samples had the same time-temperature histories it appears that carbon loss from the tubing to the flowing lithium affected more of the wall thickness for the 0.013-in. thick tubing than it did for the 0.025-in. thick tubing. Data for both tests shown in Table 5 indicate that the loss of carbon in the tubing exposed to flowing lithium resulted in an additional decrease in creep-rupture beyond that due to overaging (the tubing exposed to static argon).

For the 0.013-in. thick tubing exposed to flowing lithium it is not certain that the tubing received the 1 h solution heat treatment at 1593°C plus the 2 h anneal at 1204°C. If the material was tested in the cold worked condition the datum contained in Table 5 indicate that cold work, or cold work plus elevated temperature thermal exposure, is very detrimental to the creep-rupture properties of PWC-11 since material in this condition had the shortest rupture life of any of the material tested. The cold work not only results in higher creep rates but also could contribute to accelerated overaging.

It has been reported that "extended exposure (undefined time) of this alloy to a temperature of 1204°C (1477 K) produced coalescence and growth of zirconium-carbon-nitrogen interstitial phase" and "that the microstructure produced in samples with stress and without stress were noticeably different" (Gregory and Rowe 1961). It is not uncommon in alloys such as

PWC-11 for the applied stresses imposed during the creep testing to enhance aging. Some quantitative information on the effects of thermal aging, with and without stress on this alloy has been reported in the Soviet literature (Lyutyi et.al 1978, Girevosh et.al 1973, Maksimovich et.al 1978, Kissil et.al 1978, and Avtonomov, et.al 1978). These results are in good agreement with the above observations and comments.

If there is an interest in using this alloy for SP-100 applications, the following issues must be addressed: (1) the effects of thermal aging on the tensile and short-term (10^2—10^3 h) creep properties should be measured at temperatures relevant to proposed applications, (2) longer-term ($\sim 10^4$ h) creep tests should be initiated to provide 2 to 3% creep strain at relevant operating temperatures in order to provide more reliable estimates of the creep properties, and (3) measurement of the high-temperature creep properties, especially ductility, and low-temperature ductility and fracture toughness of weldments is required to determine whether carbide redistribution and precipitation during the welding process has resulted in low ductility and/or toughness values.

T-111 (Ta-8% W-2% Hf)

Table 6 contains pertinent information on the data obtained for T-111. A plot of Larson-Miller parameter versus applied stress for T-111 is shown in Fig. 8. The curve in Fig. 8 for time to 1% creep strain data is based only on data for time (in hours) to 1% creep strain and only for samples that were annealed at 1649°C (1922 K) for 1 h prior to testing (Sheffler 1970, Sheffler and Ebert 1973). This material should have a recrystallized

Table 6. Available data for creep tests on T-111

Reference	No. of tests	Tests used for reference data base	Total test time (h)	Total time of tests used (h)	Longest test (h)	Longest test time used (h)
Sheffler and Ebert, 1973	48	34[a]	304,161[b]	208,546	38,129	22,476
Stephenson, 1967	13[c]	0	4,925	0	1,700	0
Titran and Hall, 1962	2	0[a]	675	0	350	0
TOTALS	63	34	309,761	208,546		

[a]Time to 1% creep strain not reported.

[b]Time to rupture or discontinuance of test.

[c]All samples were tested in cold worked condition.

Table 7. Available data for creep tests on ASTAR-811C

Reference	No. of tests	Tests used for reference data base	Total test time (h)	Total time of tests used (h)	Longest test (h)	Longest test time used (h)
Sheffler and Ebert, 1973	23[a]	22[b]	154,926[c]	124,756	23,692	23,692
Sheffler and Ebert, 1973	29[d]	24[b]	86,095	56,937	13,413	8,877
Sheffler and Ebert, 1973	18	8[e]	11,958	7,661	1,945	1,777
TOTALS	70	54	252,979	189,354		

[a]Annealed 0.5 h at 1982°C.

[b]Tests contain no data on times to 1% creep strain, only extrapolated values.

[c]Time to rupture or discontinuance of test.

[d]Annealed 1 h at 1649°C.

[e]Several different TMTs tested.

structure with equiaxed grains and an optimum, or near optimum, grain size. This TMT should provide the most representative creep strength for the T-111 alloy. Data for 13 tests on material tested in the cold worked condition are not included in Fig. 8 (Stephenson 1967). Data for two tests on material annealed for 1 h at 1427°C (1700 K) fall from about 25 to 45% below the curve shown in Fig. 8 (Titran and Hall 1962).

As shown in Fig. 8 creep strength decreases slowly with increasing temperature over the range from 800 to about 1125°C (1073—1400 K) and then decrease rapidly with increasing temperature above about 1125°C (1400 K).

ASTAR-811C (Ta-8% W-1% Re-0.7% Hf-0.025% C)

Table 7 contains the pertinent information on the data obtained for ASTAR-811C. Figure 9 is a plot of Larson-Miller parameter versus applied stress for ASTAR-811C. All data are for measured time (in hours) to 1% creep strain. Two sets of data are emphasized. Twenty-four tests are on material annealed for 1 h at 1649°C (1922 K) which has a small grain size and 20 tests are on material annealed for 0.5 h at 1982°C (2255 K) which has a larger grain size (Sheffler and Ebert 1973). The larger grain size material has higher creep strength at the higher test temperatures due to the reduced grain boundary area. Note in Fig. 9, especially for the larger grain size material, that there is a pronounced inflection in the Larson-Miller parameter versus applied stress curve at approximately 1400°C (1675 K). Above about 1400°C (1675 K) the decrease in stress to produce 1% creep strain with increasing temperature is less than for temperatures between approximately 1200 and 1400°C (~1475 and 1675 K). This phenomenon is due to grain growth

Table 8. Predicted stress[a] to produce 1% creep strain in seven years

Alloy	Temperature (K)				
	1300	1350	1400	1450	1500
			Stress (MPa)		
Nb-1% Zr	13	9	6	-	-
PWC-11	45	28	16	-	-
T-111	85	63	40	18	8 (at 1485)
ASTAR-811C	180	150	125	95	70

[a]Predicted from Larson-Miller parameters based on shorter-term tests.

approximately 2% can be tolerated (Fig. 11). If the strain is maintained for 10 minutes at the tensile strain limit of the cycle, the maximum strain range that can be tolerated is reduced to 1%; if the strain is maintained for 10 h at the maximum strain limit, the allowable strain range is reduced to only 0.4%.

For SP-100 the reactor is intended to produce electrical power for 7 years during a 10-year lifetime. It is expected that the reactor system will be subjected to some type of cyclic change in load with long periods under creep loading at high temperature when the reactor is producing power. Very long periods (few to several months) at low to very low strains (0.2–0.5%) have been observed to be extremely detrimental to the load-carrying capabilities of types 304 (Maiya 1981) and 316 stainless steel (Hales 1980), 2 1/4 Cr-1 Mo steel (Brinkman 1983), and alloy 718 (Thakker and Cowles 1983). In fact, current analyses indicate that failure in components of liquid metal fast breeder reactors (LMFBRs) will most probably be due to creep-fatigue interaction due to low stresses during long operating time at temperatures >0.3 T_m and the cyclic nature of electric power plant operation. No information exists on the effects of creep-fatigue loading for the refractory metal alloys that are under consideration for use in SP-100; however, considering the reduction in load carrying capabilities that has been observed for the afore mentioned alloys, creep-fatigue testing of these refractory metal alloys should be included in the development of the engineering data base for refractory metal alloys.

High-Cycle Fatigue

For the space power systems that are based on the Stirling engine concept, the load carrying capabilities under high-cycle fatigue ($>10^8$ cycles) loading must be evaluated. For at least one of the proposed space reactor power systems the Stirling engines are to be operated at a frequency of approximately 100 Hz. For seven years of operation these engines and their components will be subjected to approximately 2.2×10^{10} cycles of low strain range fatigue loading. No data exist on the high-cycle fatigue properties of the refractory metal alloys; however, Fig. 12 shows cycles to failure as a function of strain range for type 316 stainless steel tested at 593°C (866 K) (Raske 1984). The curves in Fig. 12 are for fully reversed fatigue loading in which the sample is subjected to equal strain in tension and compression. The materials in the Stirling engines will be subjected to only the tensile portion of this loading. During such loading cycles mean tensile stresses will most probably be developed and tensile mean stresses reduce fatigue and compressive mean stresses increase fatigue life at a given amplitude of loading (Sandor 1972, Landgraf 1970). Therefore, evaluation of the load carrying capability of materials for space nuclear power concepts that employ Stirling engines must include a consideration of the high-cycle fatigue properties of the materials of which the engines are constructed.

Summary

- Data acquisition, evaluation, and analysis of the creep properties of refractory metal alloys that are candidates for structural and/or fuel cladding applications in space nuclear power systems has been initiated.

- To date evaluation of the creep strength of Nb-1% Zr, PWC-11, T-111, and ASTAR-811C using the Larson-Miller parameter has been conducted; evaluation using the Manson-Haferd and Dorn parameters is in progress.

- The available data for Nb-1% Zr, T-111, and ASTAR-811C are adequate to enable the creep properties of these alloys to be used in conceptual design studies. ASTAR-811C has the highest creep strength of the alloys evaluated and the long-term creep strength exceeds present design requirements for SP-100. The creep strengths of Nb-1% Zr and T-111 are adequate for SP-100 design requirements to approximately 1300 and 1400 K, respectively.

- For PWC-11 the available data are not sufficient to enable the creep properties of this alloy to be used in conceptual design studies. Available data indicate that PWC-11 is thermally unstable (overages) for long-term use in the proposed temperature range of operation for SP-100. The effects of aging on the elevated temperature strength properties of this alloy must be determined before any additional work is conducted on this alloy.

- For Nb-1% Zr the creep strength of weldments is similar to that of base metal.

- The thermomechanical treatment that results in maximum long-term creep strength has not been optimized for any of these four alloys.

- Creep-fatigue loading must be considered in establishing load carrying capability for space nuclear power systems; there are no creep-fatigue data for refractory metal alloys.

- For systems that utilize Stirling engines knowledge of the high-cycle fatigue properties of the structural materials in the engines is required to quantitatively evaluate load carrying capability.

ACKNOWLEDGEMENT

The author wishes to acknowledge the valuable discussions with C. R. Brinkman and D. T. Raske on fatigue and creep-fatigue loading, to R. H. Cooper and R. W. Swindeman for helpful discussions and their review of the manuscript, and to Donna Amburn for preparing the figures. Special thanks are extended to E. E. Hoffman who provided many of the references that contained the data utilized herein and/or provided helpful information concerning the data contained in the references.

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Figure Captions

Fig. 1. Plot of Larson-Miller parameter versus applied stress to produce 1% creep strain for Nb-1% Zr.

Fig. 2. Plot of Larson-Miller parameter versus applied stress to produce 1% creep strain for shorter-term tests of Nb-1% Zr in lithium.

Fig. 3. Plot of Larson-Miller parameter versus applied stress to produce 1% creep strain for Nb-1% Zr tested in vacuum.

Fig. 4. Plot of Larson-Miller parameter as a function of applied stress to produce 1% creep strain for PWC-11.

Fig. 5. Effect of carbon content on the stress to produce 1% creep strain versus Larson-Miller parameter for PWC-11.

Fig. 6. Larson-Miller plot for PWC-11 to illustrate the difference in applied stress required to produce 1% creep strain as predicted from minimum-creep-rate data and from measured time to 1% creep strain data.

Fig. 7. Plot of minimum-creep-rate as a function of applied stress for PWC-11 at several test temperatures.

Fig. 8. Plot of Larson-Miller parameter versus applied stress to produce 1% creep strain for T-111.

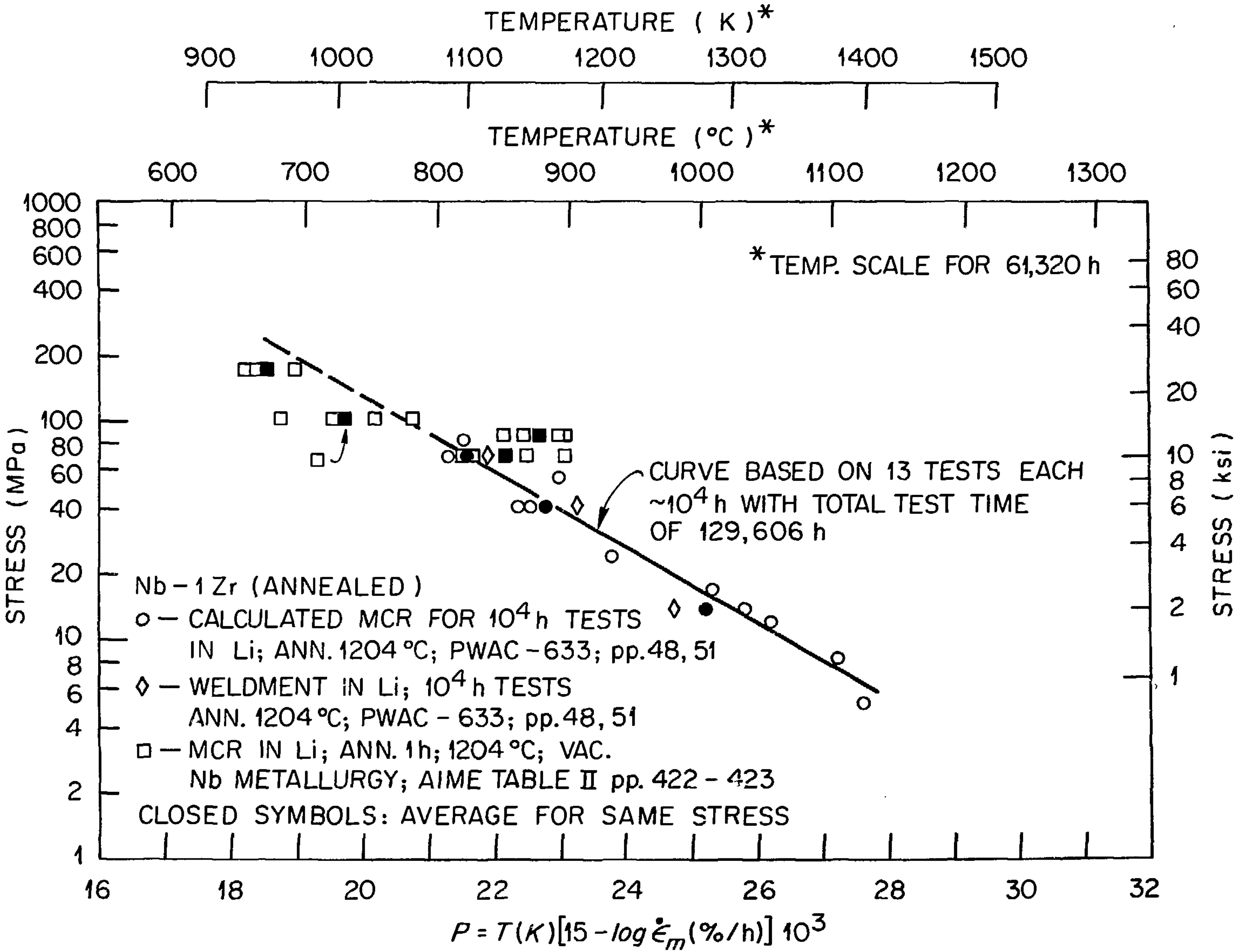
Fig. 9. Plot of Larson-Miller parameter versus applied stress to produce 1% creep strain in ASTAR-811C.

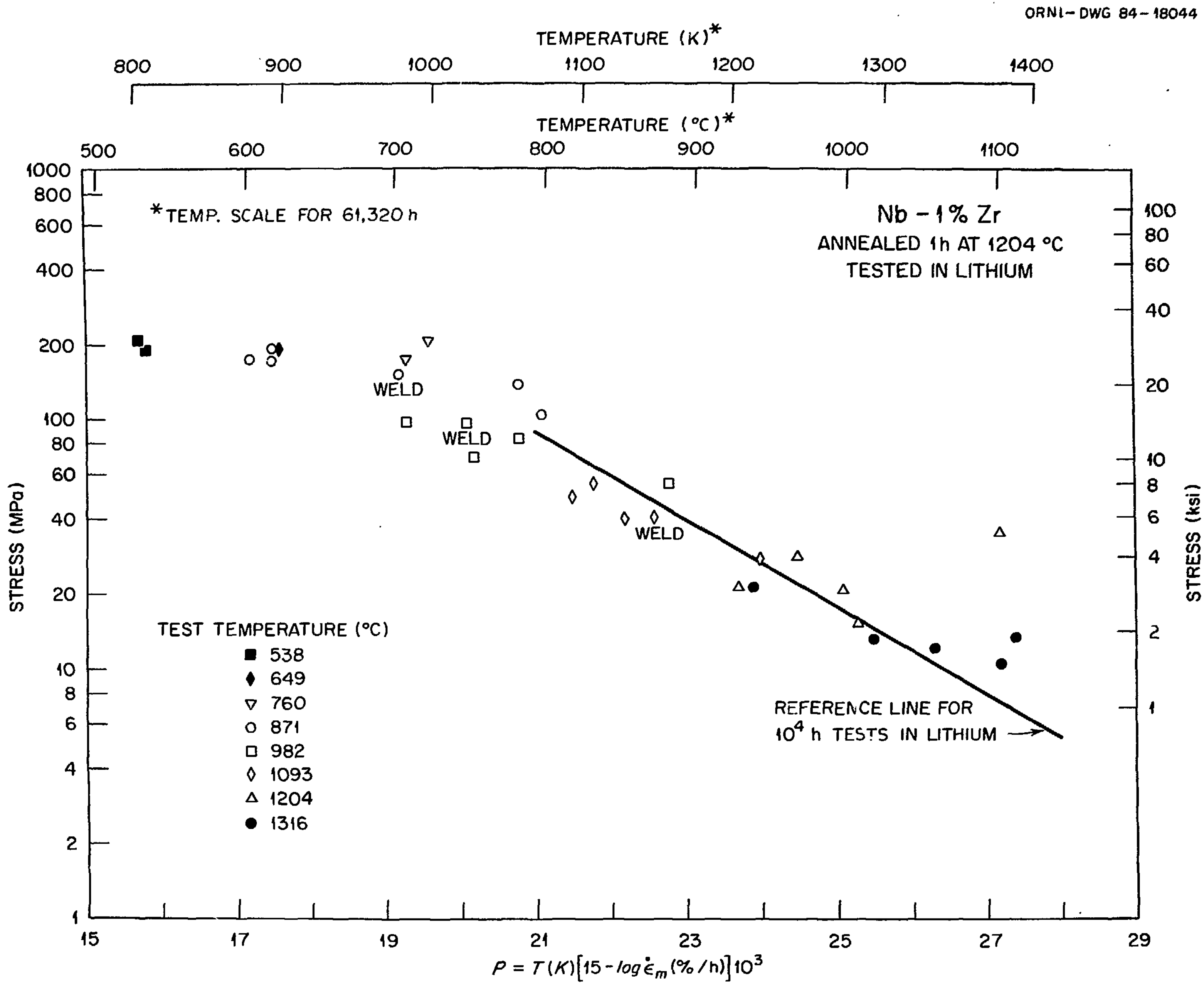
Fig. 10. Plot of applied stress required to produce 1% creep strain in 7 years as a function of temperature for Nb-1% Zr, PWC-11, T-111, and ASTAR-811C. Temperature and stress values shown are predictions from shorter-term tests using the Larson-Miller parameter.

Fig. 11. Plot of cycles to failure as a function of applied strain range for type 304 stainless steel to illustrate how creep deformation during elevated temperature portion of loading under slowly cycled operation reduces load carry capability.

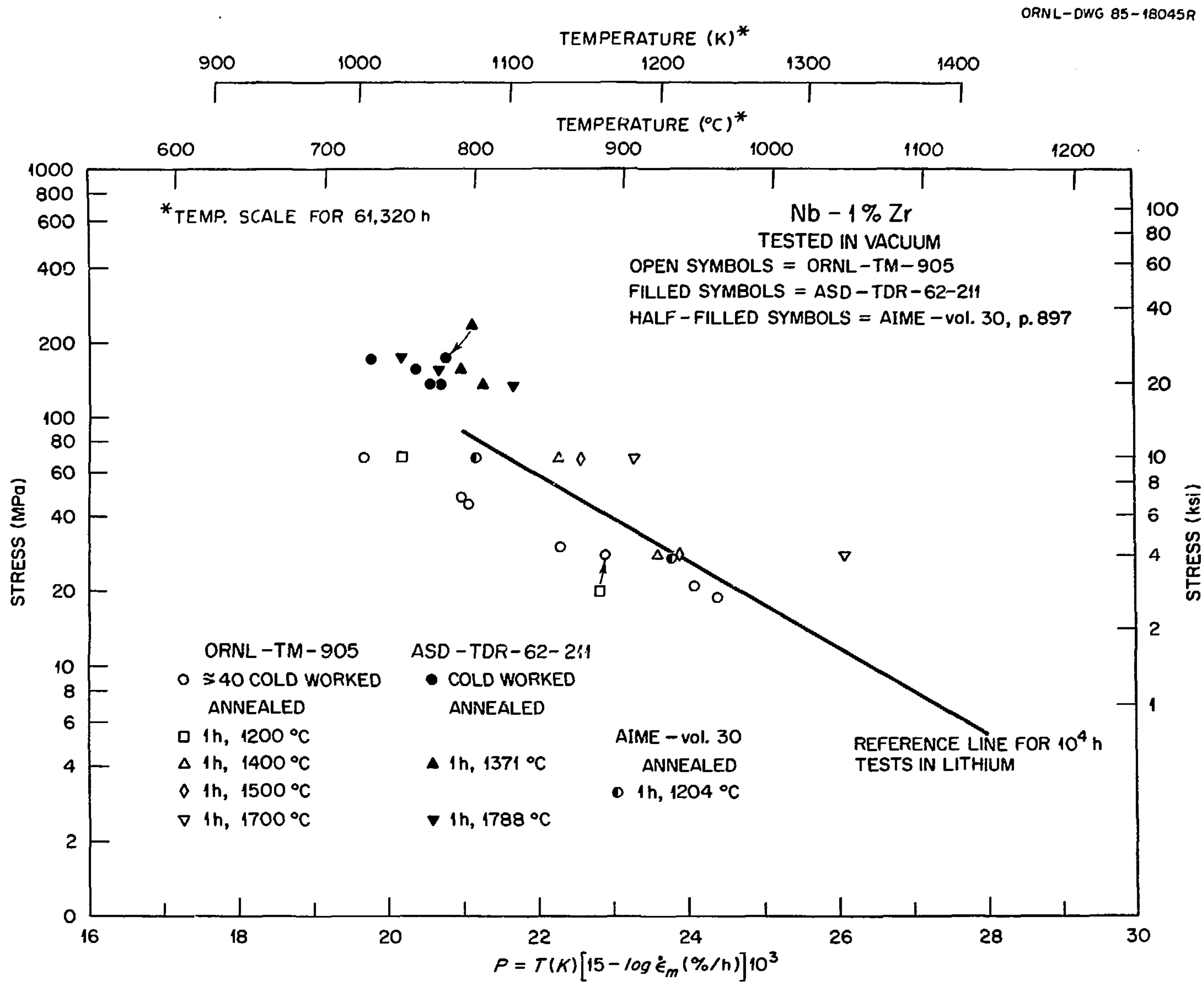
Fig. 12. Plot of cycles to failure as a function of applied strain range for type 316 stainless steel to illustrate that high-cycle fatigue may effect the load carrying capability for systems that utilize Stirling engines that operate at approximately 100 Hz.

ORNL-DWG 84-14735R



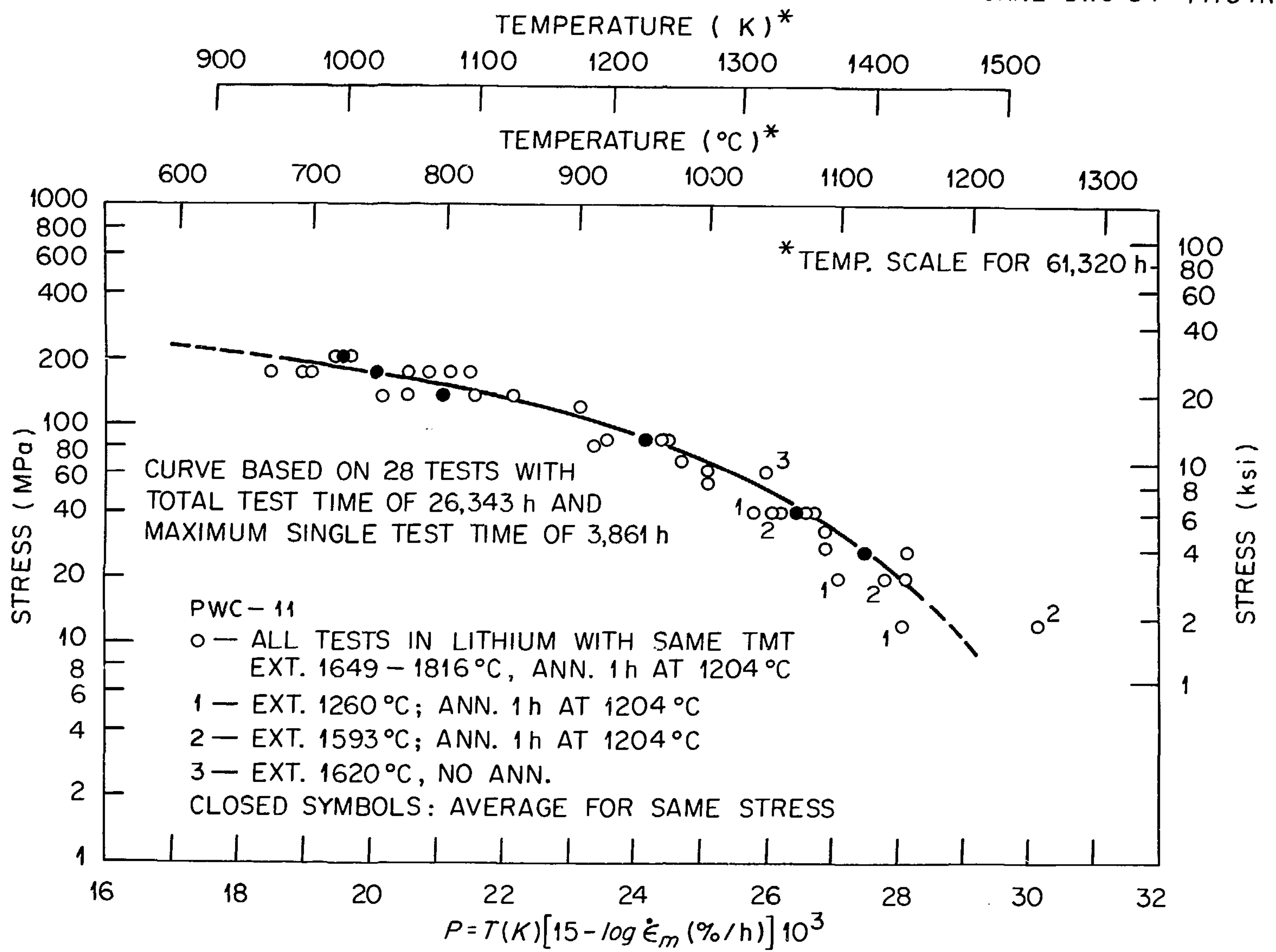


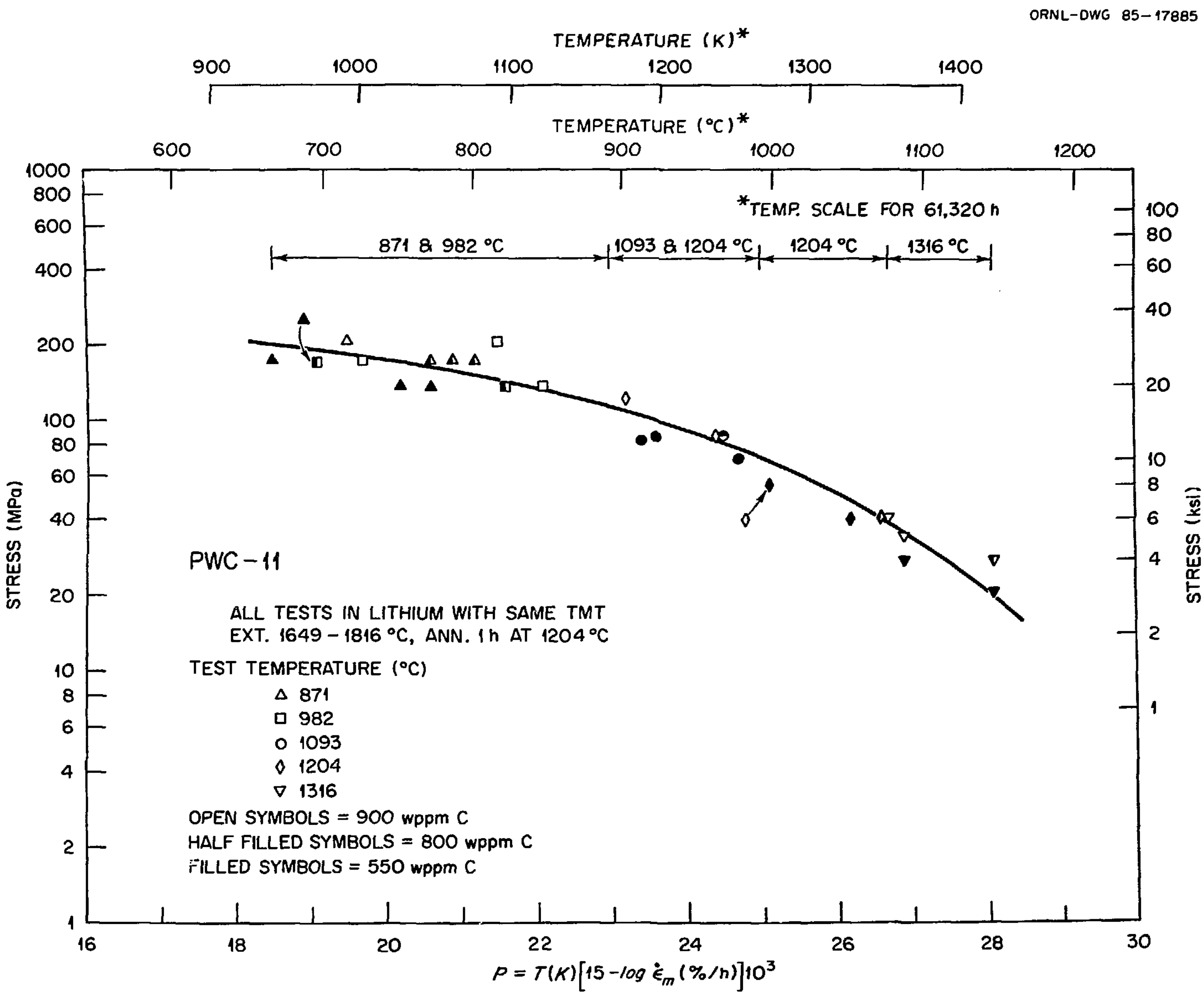
ORNL-DWG 84-18044
TEMPERATURE (K)*
800
900
1000
1100
1200
1300
1400
TEMPERATURE (°C)*
500
600
700
800
900
1000
1100
*TEMP. SCALE FOR 61,320 h
Nb – 1% Zr
ANNEALED 1h AT 1204 °C
TESTED IN LITHIUM
WELD
WELD
WELD
STRESS (MPa)
STRESS (ksi)
TEST TEMPERATURE (°C)
538
649
760
871
982
1093
1204
1316
REFERENCE LINE FOR
10⁴ h TESTS IN LITHIUM
$P = T(K)[15 - \log \dot{\epsilon}_m (\%/h)]10^3$



ORNL-DWG 85-18045R
TEMPERATURE (K)*
900 1000 1100 1200 1300 1400
TEMPERATURE (°C)*
600 700 800 900 1000 1100 1200
*TEMP. SCALE FOR 61,320 h
Nb - 1 % Zr
TESTED IN VACUUM
OPEN SYMBOLS = ORNL-TM-905
FILLED SYMBOLS = ASD-TDR-62-211
HALF-FILLED SYMBOLS = AIME-vol. 30, p. 897
STRESS (MPa)
1000 800 600 400 200 100 80 60 40 20 10 8 6 4 2 0
STRESS (ksi)
100 80 60 40 20 10 8 6 4 2 1
ORNL-TM-905
≲40 COLD WORKED
ANNEALED
1h, 1200 °C
1h, 1400 °C
1h, 1500 °C
1h, 1700 °C
ASD-TDR-62-211
COLD WORKED
ANNEALED
1h, 1371 °C
1h, 1788 °C
AIME-vol. 30
ANNEALED
1h, 1204 °C
REFERENCE LINE FOR 10⁴ h
TESTS IN LITHIUM
16 18 20 22 24 26 28 30
P = T(K)[15 - log ε̇m (%/h)]10³

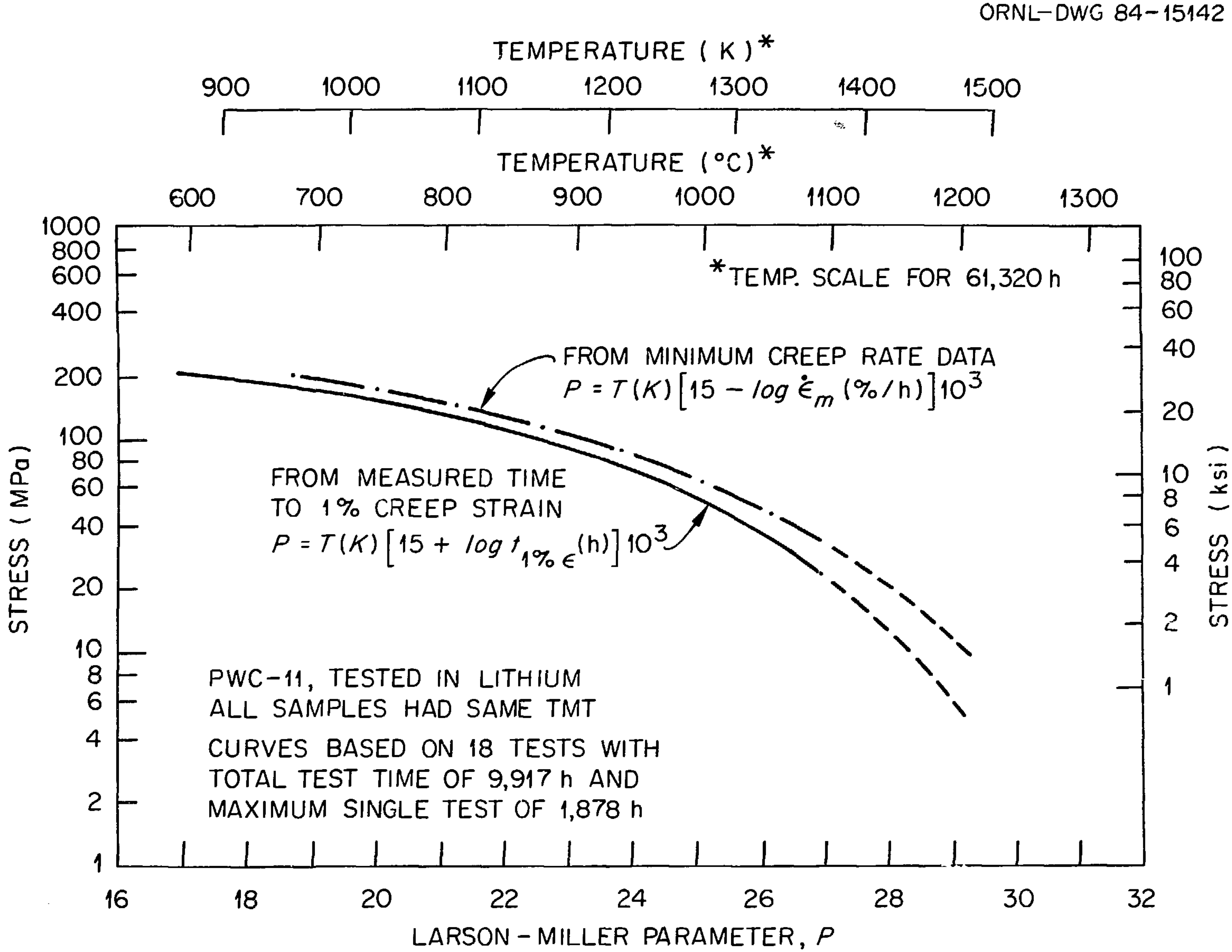
ORNL-DWG 84-14734R



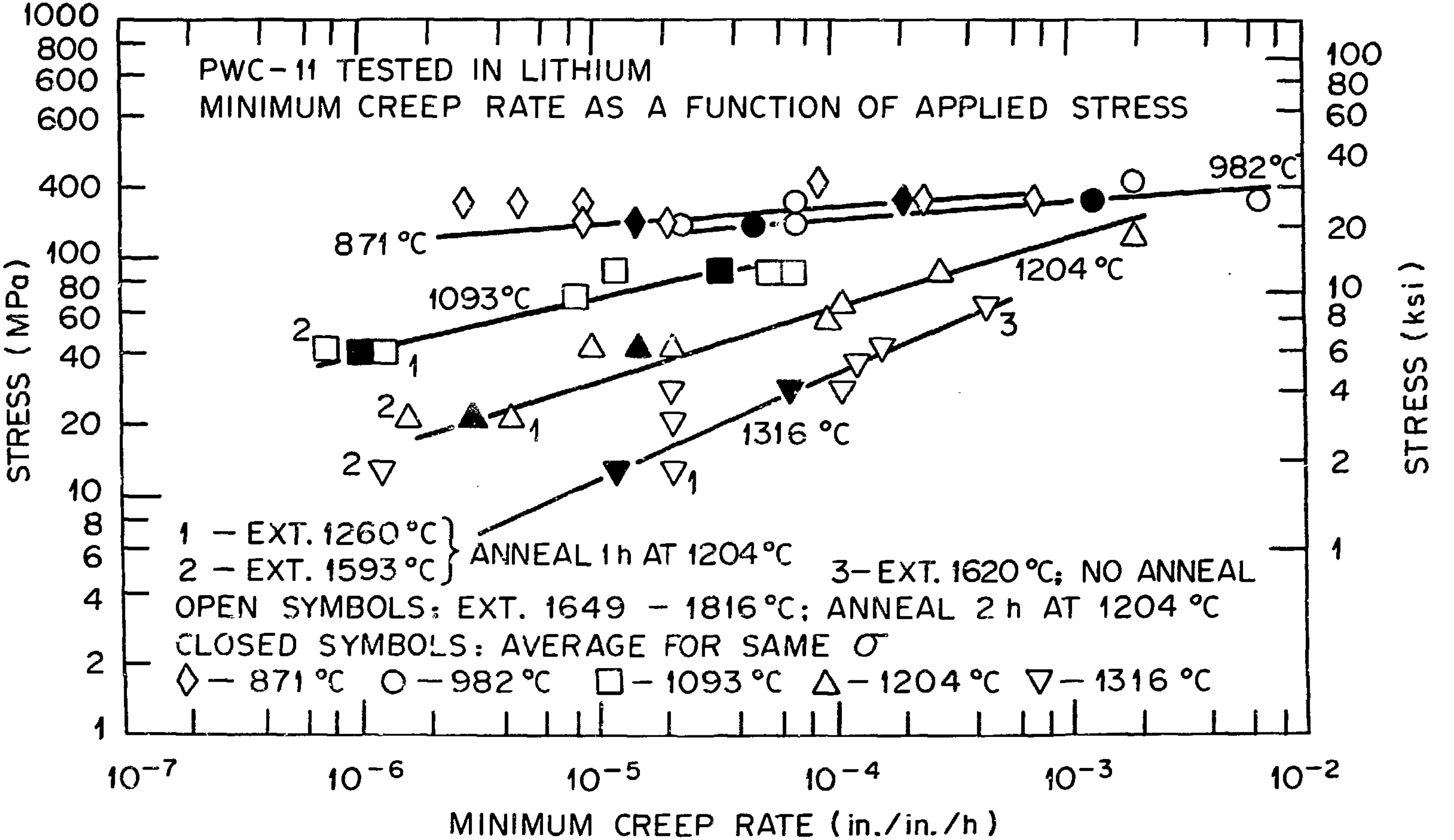


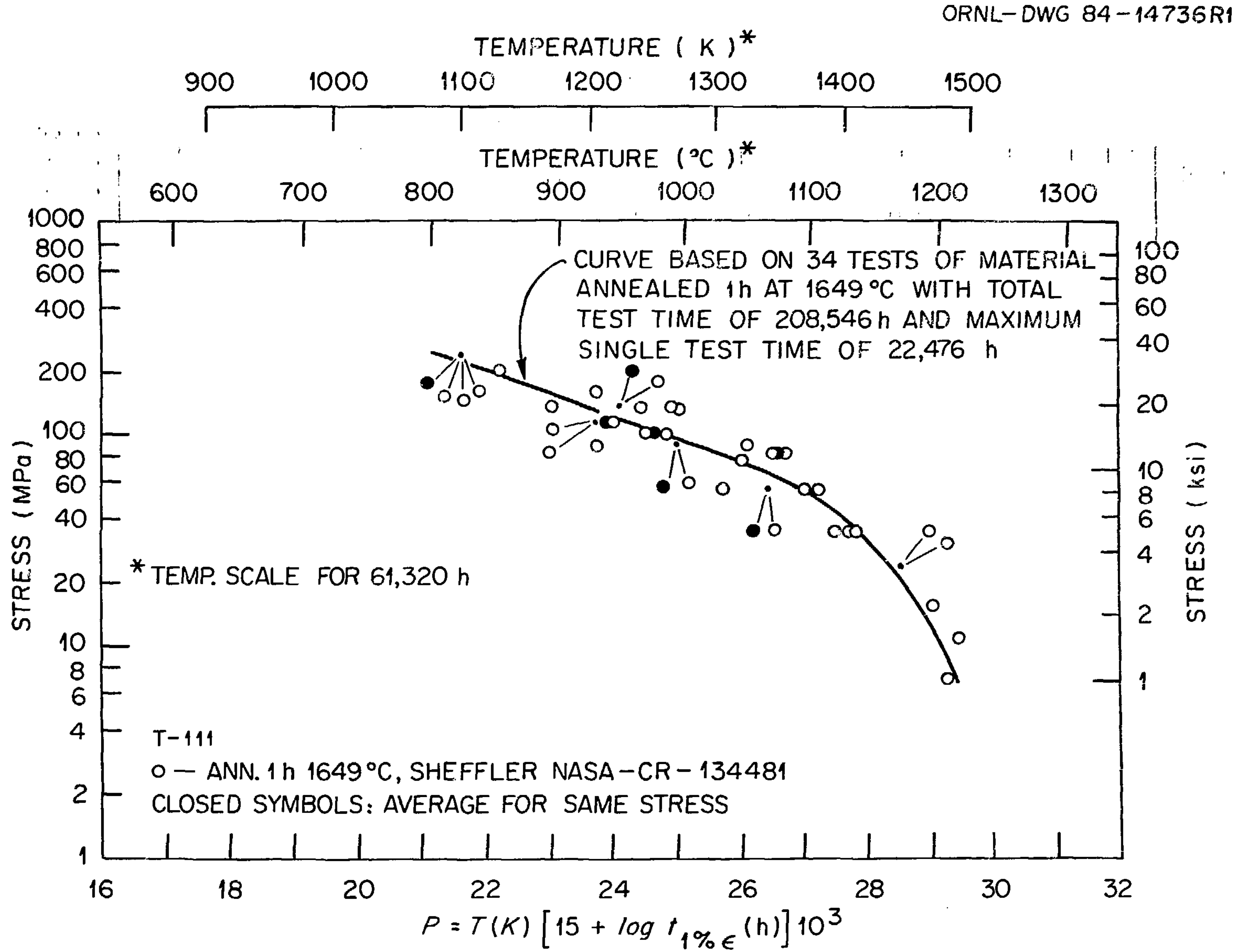
ORNL-DWG 85-17885
TEMPERATURE (K)*
900
1000
1100
1200
1300
1400
TEMPERATURE (°C)*
600
700
800
900
1000
1100
1200
*TEMP. SCALE FOR 61,320 h
871 & 982 °C
1093 & 1204 °C
1204 °C
1316 °C
STRESS (MPa)
STRESS (ksi)
PWC-11
ALL TESTS IN LITHIUM WITH SAME TMT
EXT. 1649-1816 °C, ANN. 1 h AT 1204 °C
TEST TEMPERATURE (°C)
871
982
1093
1204
1316
OPEN SYMBOLS = 900 wppm C
HALF FILLED SYMBOLS = 800 wppm C
FILLED SYMBOLS = 550 wppm C
P = T(K)[15 - log ε̇m (%/h)]10³

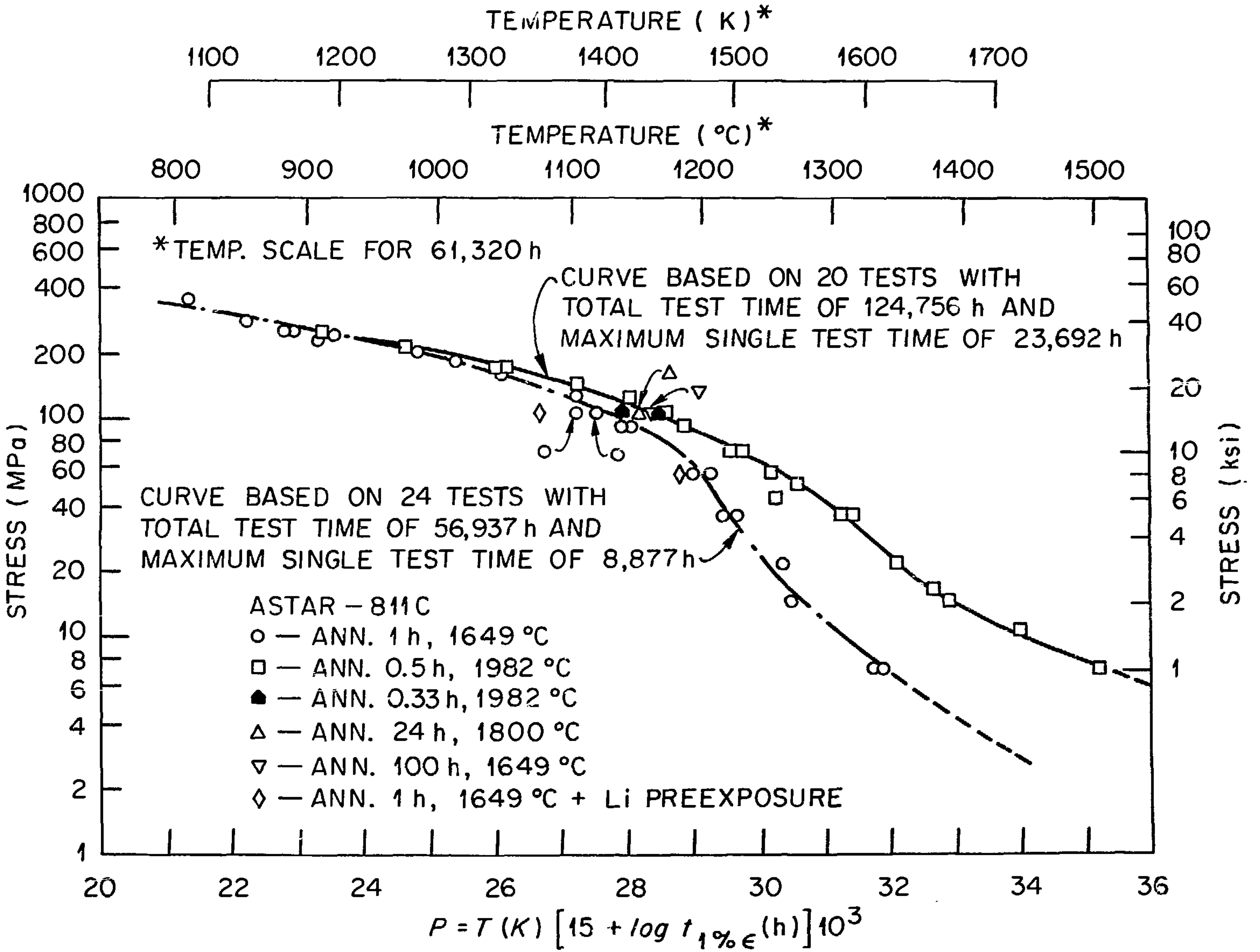
LARSON-MILLER PARAMETER VERSUS APPLIED STRESS FOR PWC-11

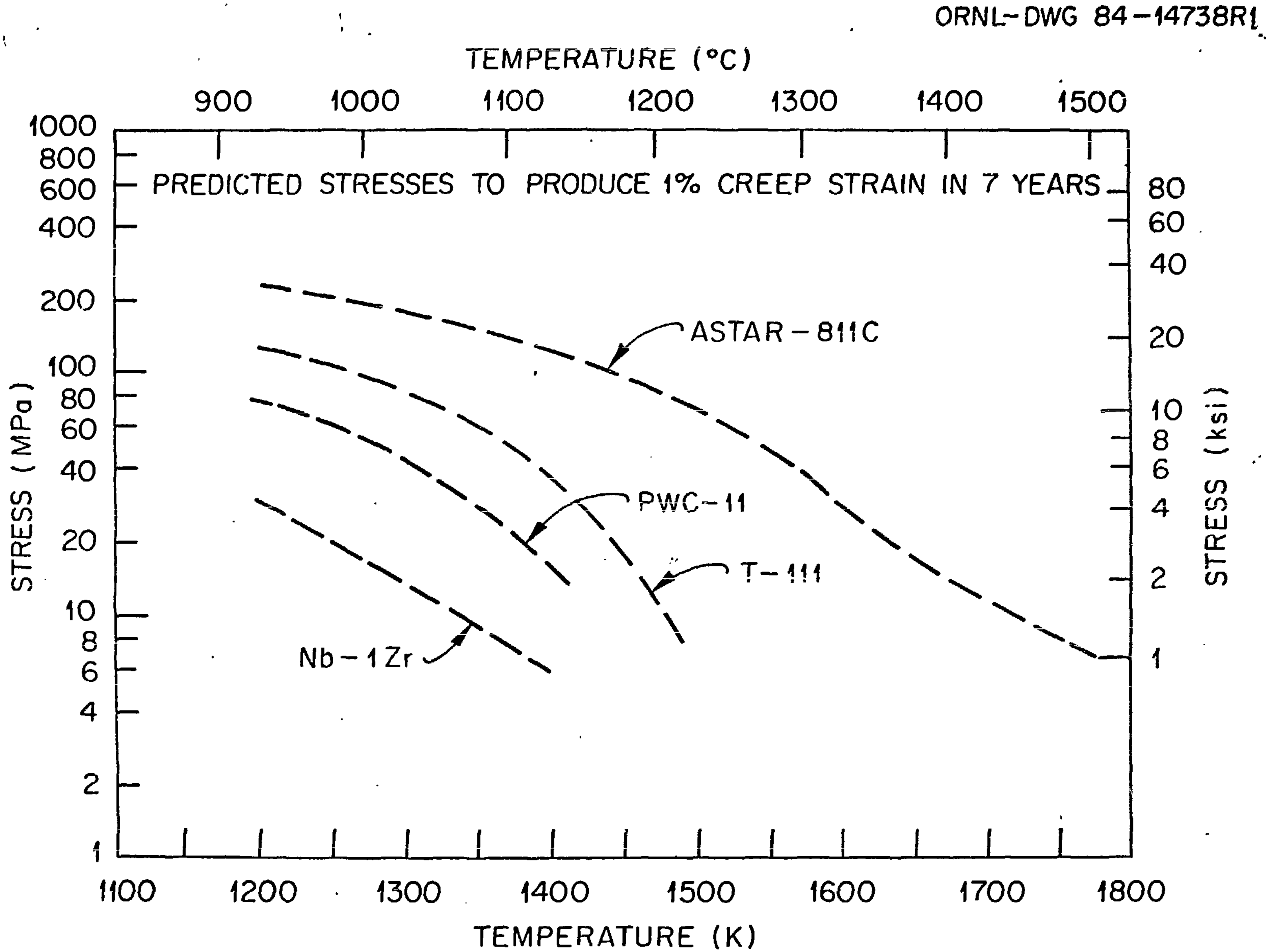


ORNL-DWG 84-13017R



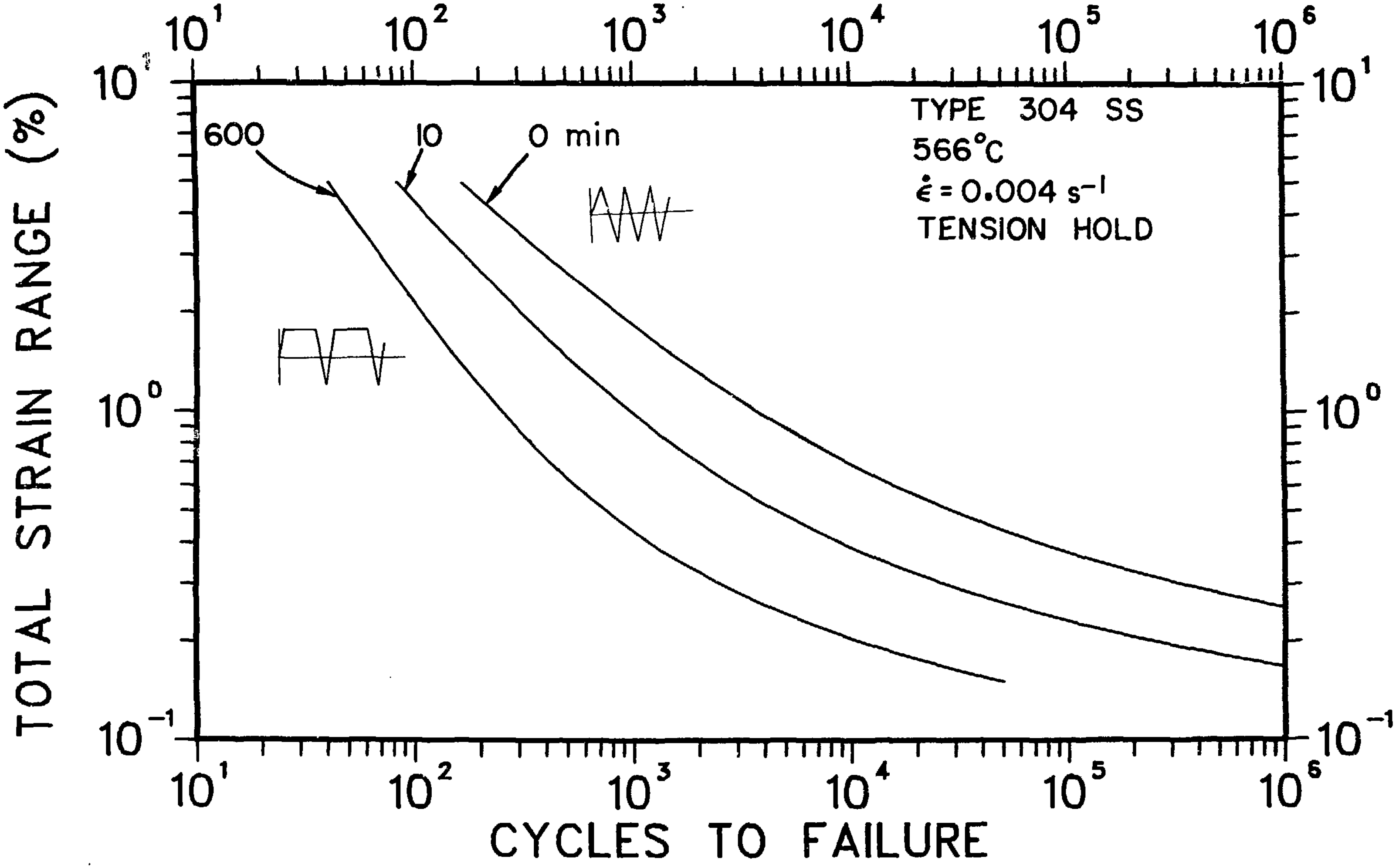

ORNL–DWG 84–14736R1
TEMPERATURE (K)*
900 1000 1100 1200 1300 1400 1500
TEMPERATURE (°C)*
600 700 800 900 1000 1100 1200 1300
CURVE BASED ON 34 TESTS OF MATERIAL ANNEALED 1 h AT 1649 °C WITH TOTAL TEST TIME OF 208,546 h AND MAXIMUM SINGLE TEST TIME OF 22,476 h
STRESS (MPa)
1000 800 600 400 200 100 80 60 40 20 10 8 6 4 2 1
STRESS (ksi)
100 80 60 40 20 10 8 6 4 2 1
* TEMP. SCALE FOR 61,320 h
T-111
○ — ANN. 1 h 1649 °C, SHEFFLER NASA–CR–134481
CLOSED SYMBOLS: AVERAGE FOR SAME STRESS
16 18 20 22 24 26 28 30 32
P = T(K) [15 + log t1%ε (h)] 10³


TEMPERATURE (K)*
1100
1200
1300
1400
1500
1600
1700
TEMPERATURE (°C)*
800
900
1000
1100
1200
1300
1400
1500
*TEMP. SCALE FOR 61,320 h
CURVE BASED ON 20 TESTS WITH TOTAL TEST TIME OF 124,756 h AND MAXIMUM SINGLE TEST TIME OF 23,692 h
CURVE BASED ON 24 TESTS WITH TOTAL TEST TIME OF 56,937 h AND MAXIMUM SINGLE TEST TIME OF 8,877 h
ASTAR – 811C
○ — ANN. 1 h, 1649 °C
□ — ANN. 0.5 h, 1982 °C
● — ANN. 0.33 h, 1982 °C
△ — ANN. 24 h, 1800 °C
▽ — ANN. 100 h, 1649 °C
◇ — ANN. 1 h, 1649 °C + Li PREEXPOSURE
STRESS (MPa)
1000
800
600
400
200
100
80
60
40
20
10
8
6
4
2
1
STRESS (ksi)
100
80
60
40
20
10
8
6
4
2
1
20
22
24
26
28
30
32
34
36
P = T (K) [15 + log t 1% ε (h)] 10^3

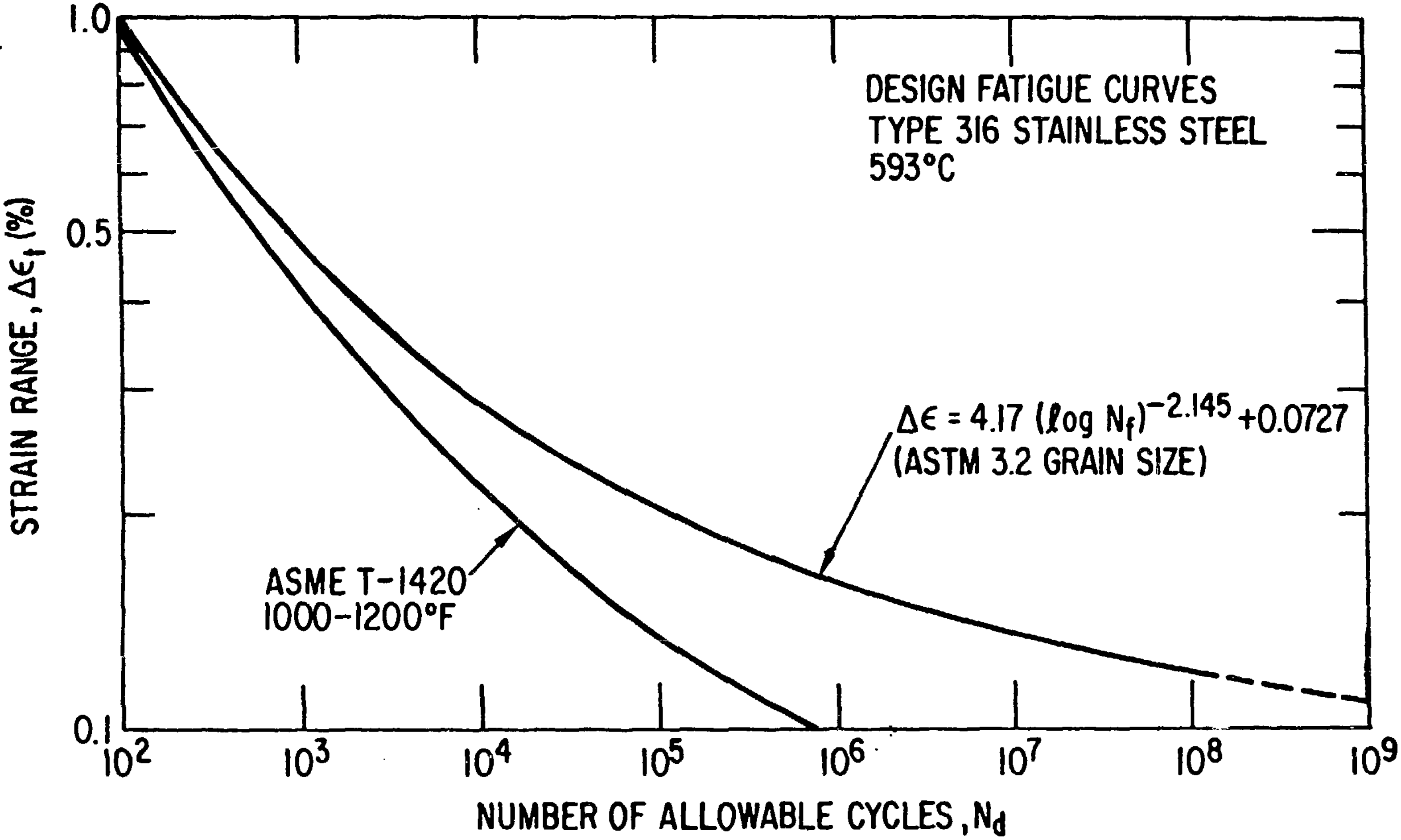


PLOT OF APPLIED STRESS REQUIRED TO PRODUCE 1% CREEP STRAIN IN SEVEN YEARS AS A FUNCTION OF TEMPERATURE FOR Nb-1% Zr PWC-11, T-111, AND ASTAR-811C. TEMPERATURE AND STRESS VALUES SHOWN ARE PREDICTIONS FROM SHORTER TERM TESTS USING THE LARSON-MILLER PARAMETER.

ORNL-DWG 83-17341



PLOT OF CYCLES TO FAILURE AS A FUNCTION OF APPLIED STRAIN RANGE FOR TYPE 304 STAINLESS STEEL TO ILLUSTRATE HOW CREEP DEFORMATION DURING ELEVATED TEMPERATURE PORTION OF LOADING UNDER SLOWLY CYCLED OPERATION REDUCES LOAD CARRY CAPABILITY



PLOT OF CYCLES TO FAILURE AS A FUNCTION OF APPLIED STRAIN RANGE FOR TYPE 316 STAINLESS STEEL TO ILLUSTRATE THAT HIGH-CYCLE FATIGUE MAY AFFECT THE LOAD CARRYING CAPABILITY FOR SYSTEMS THAT UTILIZE STIRLING ENGINES THAT OPERATE AT APPROXIMATELY 100 HERTZ

DISCLAIMER