FEB 9 1965

EURAE- 1299
ATL-A-144
Joint U.S.-Euratom Research
and Development Program

American Radiator & Standard Sanitary Corporation
Advanced Technology Laboratories Division

MASTER

Investigation of Thermal-Stress-Fatigue Behavior of Stainless Steels

Prepared under
AEC Contract AT(04-3)-250
Project Agreement No. 11
for the
Joint U.S.-Euratom Research and Development Board

Mountain View, California

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ATL-A-144
Joint U. S. -Euratom Research
and Development Report
Special Distribution

INVESTIGATION OF
THERMAL-STRESS-FATIGUE BEHAVIOR
OF STAINLESS STEELS

Final Report

ATL Job 1230301

31 October 1964

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INTRODUCTION

The United States and the European Atomic Energy Community (Euratom), on May 29, and June 18, 1958, signed an agreement which provides a basis for cooperation in programs for the advancement of the peaceful applications of atomic energy. This agreement, in part, provides for the establishment of the Joint U.S. -Euratom research and development program which is aimed at reactors to be constructed in Europe under the Joint Program.

The work described in this report represents the Joint U.S. -Euratom effort which is in keeping with the spirit of cooperation in contributing to the common good by the sharing of scientific and technical information and minimizing the duplication of effort by the limited pool of technical talent available in western Europe and the United States.

FOREWORD

This final report, covering the investigation of thermal-stress-fatigue behavior of stainless steels and certain other alloys, has been prepared for the Joint U. S. -EURATOM Research and Development Board under Contract No. AT(04-3)-250, Project Agreement No. 11. The investigation was conducted by American Radiator & Standard Sanitary Corporation, Advanced Technology Laboratories (ATL) Division, during the period March 1962 through September 1964.

The final report is intended to summarize the test results discussed in the ten quarterly progress reports and the two topical reports published during this program. Certain details, such as the method of starting up a test and equipment operation, are not discussed in this report in the interest of brevity; however, the interested reader may find this information in the above-mentioned documents, which are listed in Appendix C of this report.

SUMMARY

Data on fatigue in metals have, to date, been gathered primarily by conventional test means, i. e., rapid stress cycling at constant temperatures. Such data are invariably presented as a series of constant-temperature curves of logarithmic stress versus logarithmic cycles to failure. Furthermore, many sets of such curves are necessary if testing is performed with different stress-cycling rates such that stress relaxation is allowed. In an effort to fatigue-test nuclear reactor materials under conditions similar to those that would be expected in actual practice, Advanced Technology Laboratories has been conducting a thermal-stress-fatigue program in which stress cycling was induced by thermally cycling a constrained specimen. A major departure from conventional data-presentation methods was made by plotting logarithmic plastic strain range versus logarithmic cycles to failure. It was hoped, and largely proved, that such curves would be more "independent" of test conditions than curves of stress versus cycles to failure. The major reason why data have not been presented in this new way in conventional fatigue programs is the extreme difficulty of measuring small plastic strains.

The type of research described in this report has been performed recently by a number of other investigators, with varying degrees of success. Their techniques have been studied, and from this work, coupled with trial and error, have evolved the methods used by ATL in obtaining the data presented here. Claim is not made, however, to having developed the ultimate technique.

This report summarizes the results of thermal-stress-fatigue testing on types 304 and 304-L austenitic stainless steels; type 403 martensitic stainless steel; types A302-B and A387-D (Croloy 2-1/4) steels; Incoloy 800, and Zircaloy-2 (Zr-2). The thermal-stress-fatigue behavior of these alloys was determined under a wide variety of test conditions. The scope of research was intended to be sufficient to insure that the data generated would be applicable by the engineering community in general and the nuclear community in particular.

A contribution to the field of fatigue testing was made by the development of a plastic-strain-measuring device and of techniques for graphical manipulation of recorded data. It was thus possible to report fatigue life in a range less by an order of magnitude than previously reported data.

The use in this study of cyclic plastic strain instead of cyclic stress on graphs of fatigue life resulted in curves largely independent of such test conditions as mean temperature, time, and strain rate. The results will hence be of value, with little or no extrapolation, to anyone required to estimate thermal-stress-fatigue life. Furthermore, it has been shown that the conventional (strain-cycling, constant-temperature) fatigue life often agrees with the thermal-stress-fatigue life on a strain basis of comparison, and in these cases the conventional fatigue test can be used as a substitute for the thermal-stress-fatigue test.

Since the majority of fatigue results from earlier work are plotted on a stress basis, the test results obtained in this program are also plotted on this basis. Such plots are shown to be of limited value because of their strong dependence on test conditions. In this connection, an argument is presented for the paramount importance of the ability of a material to withstand cyclic plastic strain, as opposed to the importance of material strength.

BACKGROUND

A. History

The phenomenon of fatigue fracture is utilized every day to aid man, usually in such menial tasks as breaking in two a paper clip or piece of wire. On the other hand, the occurrence of fatigue failure has plagued man throughout history by causing loss of structures and lives in a catastrophic manner. Quite naturally man began studying this phenomenon by cycling some sort of specimen through a fixed strain or stress amplitude and noting how many cycles could be withstood before failure, at constant temperature of course. It often happens, however, that temperature varies cyclically in a structure, such as in an airplane, pressure vessel, or nuclear reactor, and one then questions the validity of applying constant-temperature, fatigue-test results to applications involving cyclic temperature changes. The advent of thermal-stress-fatigue testing represented the initial attempt to study fatigue under such cyclic temperature conditions as mentioned above.

The early results of thermal-stress-fatigue (TSF) tests varied greatly from conventional stress or strain-cycling-fatigue (SCF) tests. TSF tests are inherently an order of magnitude more difficult to perform, and to obtain accurate results from, than SCF tests. Hence, some excuses can be made for these early results, even though some reviews of the subject have been more than critical of them.

Further refinements in test procedures and in methods of presenting test results have greatly improved the field of TSF testing. The conclusion that TSF and SCF test results are identical over a wide range of conditions is a result of these refinements, and TSF testing has, so to speak, finally come "into its own." It must be recognized, however, that if it were not for the early investigations the subject of TSF would still be vastly complex and unclear.

B. Present Work

The primary purpose of this research program was to obtain valid TSF results that could be used easily and confidently by designers. Clearly, it was desirable to vary test conditions widely in order to assure the user of this data of its applicability to his problem.

The scope of this program was chosen to include a metallurgical study into the causes and manifestations of TSF resulting from a wide range of test variables imposed on selected stainless steels. As the program progressed, it was decided to include reactor alloys in general.

TEST EQUIPMENT, MATERIALS, AND PROCEDURE

A. Apparatus

Schematic drawings of the TSF and SCF apparatus are shown in Figures 1 and 2, respectively. These fixtures permit fatigue strain to be applied by a specimen temperature change in the case of TSF and by an externally induced mechanical strain in the case of SCF. The procedure for starting a fatigue test has been amply described in earlier work conducted at ATL^{1*} and in a topical report on this program.² Since these machines are instrumented for measuring load, strain, and temperature, it was found desirable to build an instrument control and recording console to permit all six machines (four TSF and two SCF) to be monitored. This panel is shown in Figure 3. The panel switch layout enabled a technician to take all required test data.

B. Materials

The compositions and as-received condition of the materials tested are given in Table I.^{**} The usual starting form from which test specimens were machined was $1\frac{1}{4}$ -inch-round bar. The two exceptions to this starting form were the A387-D (Croloy 2-1/4) steel and the Zircaloy-2. Both of these materials were received in tube form, requiring that end plates be welded on prior to specimen machining.

All material received from vendors was specified as being from only one heat. Thus it was assumed that no variations in chemical composition existed in test specimens of a given alloy.

* Superscripts refer to references, Appendix B.

** See Appendix D for all tables referenced.

1	Baseplate	14	Tube Connector	28	Load Cell, Baldwin SR4
2	Side Plate, Left	15	O-ring	29	Insulation, Formica
3	Test Specimen	16	Side Plate, Right	30	Screw
4	Ring, Specimen Holder	17	Power-Input Insulation	31	Washer
5	Split Ring	18	Power-Lead Bar	32	Bushing, Nylon
6	Cooling Ring	19	Screw	33	Screw
7	Washer, Brass	20	Nut	34	O-ring
8	Nut, Lock, Thin	21	Cooling-Ring Base	35	O-ring
9	Upper Plate	22	Copper Tubing	36	Screw
10	Washer	23	O-ring	37	O-ring Spacer
11	Nut, Lock, Thick	24	O-ring	38	Hold-Down Jig
12	Upper Shaft	25	Screw	39	Hold-Down Screw
	(Adjustable Screw)	26	Solid Screw	40	Cooling-Water & Power Lead
13	Gas Tube	27	Nut	41	Grip Assembly

SCHEMATIC OF TSF SPECIMEN HOLDER

FIGURE 1

SCHEMATIC OF SCF SPECIMEN HOLDER

FIGURE 2

1. OFFNER SIX-CHANNEL RECORDER
2. VARIAN SINGLE-CHANNEL RECORDER
3. L&N TEMPERATURE RECORDER
4. M-H SPEEDOMAX H RECORDER-CONTROLLER
5. BALDWIN SR-4 LOAD INDICATOR
6. COOLING-AIR (ARGON) CONTROLS
FOR TSF TESTERS
7. SWITCHES TO FEED SIGNALS
INTO VARIOUS RECORDERS
8. DAYTRONIC STRAIN INDICATOR
9. DAYTRONIC LOAD INDICATOR
10. L&N POTENTIOMETER
11. ELECTRODYNAMICS X-Y RECORDER
12. DATATRAK PROGRAMMER

CONTROLLER-RECORDER CONSOLE FOR TSF AND SCF TESTS

FIGURE 3

C. Specimen Configuration

A cross-sectional view of the specimen design used in both TSF and SCF tests is given in Figure 4. This specimen design was first used by Coffin³ and has since been used by many investigators. The temperature profile along the specimen induced by Joule heating, also included in Figure 4, represents the maximum temperature gradient in a thermal cycle. In tests with lower maximum temperatures or with a hold time at temperature extremes, the gradient was significantly less, as it was in SCF tests where temperature was of course constant. In the SCF tests, a cartridge heater placed in the axial hole of the specimen was also used at various stages of the program, with similar results. No effect on fatigue life was observable between the two heating methods, as has been suggested by others.⁴

Temperature was controlled and measured by spot-welding 28-gage Chromel/Alumel thermocouple wires to the center of the gage length. Usually these wires had no adverse effect on the specimen; occasionally, however, fatigue fractures would either start at or detour through the spot-welded regions.

Another specimen design incorporating a varying cross-sectional area was investigated; however, it was used only briefly (see following section).

FIGURE 4

D. Data Compilation

In recent years there has been a tendency to record both cyclic stress and cyclic strain in fatigue tests. Earlier, cyclic stress was the only mechanical variable recorded, perhaps primarily because this variable is easy to measure. All that is required for stress measurement is a load cell in series with the specimen. The measurement of strain, and in particular plastic strain, is more involved and is discussed below.*

The equation describing the movement, between any two points in a thermal cycle, of an extensometer attached to the specimen is

$$\ell_o' \left[\frac{\sigma_4 - \sigma_3}{E} + \epsilon_p + (T_4 - T_3) \alpha \right] = \ell_{o,4}' - \ell_{o,3}' = \Delta \ell_o' \quad , \quad (1)$$

where the number subscripts represent different points in a cycle. The first term inside the brackets is elastic mechanical movement, the second term is plastic mechanical movement, and the third is thermal movement. From this equation it is desired to extract the quantity ϵ_p , and by proper choice of points 3 and 4 it is possible to extract the plastic strain range, $\Delta \epsilon_p$. An example of a method of extracting $\Delta \epsilon_p$ from equation 1 is described briefly below and shown graphically in Figure 5. This method is based on choosing points 3 and 4 to be at zero load on heating and cooling, respectively.

To find $\Delta \epsilon_p$, X-Y recordings are obtained of the movement of ℓ_o' and the load sustained by the specimen versus temperature. The temperature difference between T_4 and T_3 (multiplied by coefficient of thermal expansion, α) corresponds to the thermal-movement term of equation 1 and is conveniently converted to a length difference, Z, by superimposing the recordings of unconstrained movement, as shown in Figure 5. The recording of constrained movement yields the length difference, Y, of ℓ_o' at zero-load points of the cycle. Since the elastic-mechanical-movement term drops out ($\sigma_4, \sigma_3 = 0$), equation 1 becomes

* See Appendix A for list of symbols.

$$\Delta\epsilon_p = \frac{Y + Z}{\ell_o'} \quad \text{or} \quad \frac{Z - Y}{\ell_o'} \quad , \quad (1a)$$

depending on the conditions of constraint.

If it happens that the recording of unconstrained movement of ℓ_o' versus temperature proves to be a loop rather than a line, as would be the case when the temperature experimentally recorded is not representative of the average, the correction is easily made by taking into account the relative displacements of the points under examination, since an error in temperature would be common to all. As may be seen, numerous sources of error are self-canceling, such as calibration errors, and variances in coefficient of thermal expansion and modulus of elasticity. Using this method with large-scale magnifications, values of $\Delta\epsilon_p$ have been determined with good reproducibility to 20 $\mu\text{in.}/\text{in.}$

There are other methods of extracting $\Delta\epsilon_p$ from equation 1. The method first used in this research program was to record both free and constrained movement of ℓ_o' versus time in a cycle. The constrained movement was then subtracted from the free movement and the remainder plotted versus load. If any plastic strain had occurred during the thermal cycle, this cross plot would be a hysteresis loop whose width was plastic movement. Dividing loop width by ℓ_o' gave plastic strain range. This early method proved to be somewhat tedious and not as accurate in low $\Delta\epsilon_p$ ranges because of the multiple cross plotting.

The later method utilizing X-Y recordings was used during all phases of the program except the initial austenitic stainless steel experiments, where the method described in the preceding paragraph was used. Both methods depend for their accuracy on a very low thermal gradient along ℓ_o' . A large thermal gradient along ℓ_o' would result in an average value of $\Delta\epsilon_p$ being extracted from equation 1. As seen from Figure 4, however, with a $\frac{1}{2}$ -inch length for ℓ_o' , a low ΔT (10 to 20°C maximum) was present along ℓ_o' and the effects of a thermal gradient were negligible.

The extensometer arrangement permitting the above measurements of movement is shown at twice actual size in Figure 6. This extensometer was designed to give a $\frac{1}{2}$ -inch length of ℓ_o' , and this length was used throughout the program. When attached to the

center of a specimen by a spring clamp arrangement of very low clamping force (~2 ounces), the extensometer does not affect the specimen. An attached extensometer is pictured in Figure 7.

In an attempt to provide a zero temperature gradient across a gage length, a specimen of varying cross-sectional (VCS) area was also used, with movements measured diametrically at the minimum diameter. This specimen design, shown in Figure 8, clearly has a diametral thermal gradient of zero at the minimum diameter; the values of $\Delta\epsilon_p$ obtained from diametral measurement are thus the true maximum values. Unfortunately, this specimen design^{*} gives erratic results in tests of relatively large $\Delta\epsilon_p$ (0.001 to 0.01 in./in.), possibly because of geometric instability in the plastic region. This design was therefore used only on tests lasting many cycles, and its use was discontinued after the initial phases of the program. The results for this specimen design are listed in Table II.

* A theoretical stress analysis was performed on this design prior to its use. See reference 5.

a. l_o' Under More than 100% Constraint, $\Delta\epsilon_{p,s} = \frac{Y + Z}{l_o'}$

b. l_o' Under Less than 100% Constraint, $\Delta\epsilon_{p,s} = \frac{Z - Y}{l_o'}$

GRAPHICAL METHOD OF OBTAINING PLASTIC STRAIN RANGE, $\Delta\epsilon_{p,s}$

FIGURE 5

TOP VIEW

SIDE VIEW

SHORT MEASUREMENT GAGE LENGTH
TSF EXTENSOMETER SCHEMATIC

FIGURE 6

TSF TEST MACHINE SHOWN WITH SPECIMEN, EXTENSOMETERS,
AND LOAD CELL IN PLACE
(Vacuum-Tight Front Cover Removed)

FIGURE 7

NEW SPECIMEN DESIGN OF VARYING CROSS-SECTIONAL AREA

FIGURE 8

TEST RESULTS

A. Test Outline

The scope of this program was undoubtedly the largest ever attempted in any thermal-stress-fatigue program to date in that large variations in test parameters and the collection of fatigue data on at least three alloy systems were called for. It was desired to vary test conditions greatly for two reasons: 1) in the hope that the resulting data would be useful under a wide variety of conditions without the need for uncertain data extrapolation, and 2) to ascertain the effects of test conditions on test results to provide some insight into the cause for fatigue fracture. The wide variation in test variables was applied mainly to the 304-L austenitic-stainless-steel phase of the program; only limited variations were attempted on the other alloys studied.

The assignment of test conditions to any given test was made by utilizing the scheme presented in Figure 9 (and included, for convenience, as Table III in Appendix D). It is seen that seven alloys were investigated instead of the three originally planned. This scheme was used consistently in preparing the tabulations of test results presented in Appendix D (see example on Figure 9 for clarification).

EXAMPLE: I.A.1.a.1.a)-1a indicates austenitic stainless steel, type 304-L, annealed (small grain), tested in air, 20-mil wall thickness, thermal-stress-fatigued with zero hold time at temperature extremes at a strain rate of 0.0005 in./in.-sec, temperature limits of cycling 500-300°C.

NOTE: ALL TEMPERATURES IN °C.

PROGRAM TEST CONDITIONS

FIGURE 9

B. Form of Results

While earlier conventional-fatigue tests reported data in the form of stress amplitude versus cycles to failure at fixed temperatures, strain rates, etc., every effort was made in this program to obtain data that could be generalized. For example, it was believed that presenting a graph of fatigue stress range at a given maximum temperature of thermal cycling would be of little value since in all probability a reactor design engineer would require fatigue data for a different temperature. Therefore, the philosophy that strain range rather than stress range is the independent variable was applied, since strain, particularly plastic strain, is a function of such parameters as stress, temperature, duration of stressing, and strain rate. It was believed that a relation of the type

$$\Delta\epsilon_p = \psi(t) \phi(T) \xi(\epsilon) \text{ etc.} \quad (2)$$

existed. Hence, if a designer had a graph of $\Delta\epsilon_p$ versus cycles to failure (N), he could substitute his reactor condition into equation 2, determine an expected $\Delta\epsilon_p$, and thus predict fatigue life, regardless of the conditions under which the fatigue data were actually obtained. In principle, a similar argument could be made for stress range ($\Delta\sigma$) versus N, but as will be shown this argument is much less effective.

A graph of $\Delta\epsilon_p$ versus N is an indication of the ability of a material to absorb plastic strain under various test conditions and after varying metallurgical histories. That is, $2N\Delta\epsilon_p$ is the total plastic strain absorbed (or withstood) to failure.* A more correct representation of this "ability to withstand" plastic strain is

$$2 \sum_{i=1}^N N_i \Delta\epsilon_{p,i} \quad , \quad (3)$$

where $\Delta\epsilon_{p,i}$ is the plastic strain range in cycle N_i . However, it is most often found that $\Delta\epsilon_p$ at half life is a good representative average of $\Delta\epsilon_{p,i}$ in any given fatigue test and hence may be substituted for $\Delta\epsilon_{p,i}$.

* Failure is defined as the onset of a drop in tensile load, a condition that corresponds closely to the initial propagation of the ultimate fatigue fracture and is readily distinguishable on the load-time recordings.

The quantity $2N\Delta\epsilon_p$ is closely connected to the total plastic strain energy absorbed during TSF testing. Since the latter quantity has been shown to be dependent on test conditions,⁶ it is logical to expect $2N\Delta\epsilon_p$ to vary also with test conditions. That is, the relation

$$2N\Delta\epsilon_p \neq \text{constant} \quad (4)$$

is expected to exist for fatigue test results.

It is well known that Coffin³ found the relation

$$N\Delta\epsilon_p = C\sqrt{N} \quad , \quad (5)$$

which suggests that $N^{k'}\Delta\epsilon_p$ is a constant. The quantity $N^{k'}\Delta\epsilon_p$ is also related to the ability to withstand plastic strain and could arbitrarily be termed the "modified ability to withstand" plastic strain. This quantity is considered important in explaining test results, as will be made evident later in this report.

C. Zircaloy-2

The test program performed on Zircaloy-2 (Zr-2) was designed to complement the previous study made by ATL on this material.¹ The earlier program was conducted at a time when measurement of small values of $\Delta\epsilon_p$ was not possible with the available equipment.

The Zr-2 data are summarized in Table IV and plotted in Figure 10, where it is seen that TSF testing in air leads to early failure. Decreasing either the maximum temperature of cycling (T_2) or the hold time at T_2 increases the TSF life. On the other hand, the same change in variables does not affect the SCF results. The fact that a specimen lasts longer when strain-cycled about a fixed temperature equal to the maximum temperature of TSF cycling is somewhat unexpected, since more oxidation would be expected in the SCF test. However, the significant difference in coefficient of thermal expansion of Zr-2 and ZrO_2 causes complex (including radial) stresses to be set up in TSF tests that would not be present in SCF tests. It must be concluded that the complex stress distribution causes more rapid initiation of a fatigue crack than would normally occur without the complex stress condition. Photomicrographs of specimens TSF- and SCF-tested at identical maximum temperatures are shown in Figure 11.

It is seen that less oxidation occurred during the TSF test than during the SCF test. No metallographic evidence of a complex stress distribution is seen in Figure 11a, as would be expected.

When TSF testing was done in argon, specimen life was increased. Figure 12 shows the results of an argon test on cold-worked Zr-2 that failed by transcrystalline fracture. There was very little oxide on this specimen. The structure shown is indicative of overheating due to fracturing and does not represent the normal cold-worked structure.

The above results strongly point to the probability that the true TSF and SCF resistance of cold-worked Zr-2 is represented by the line in Figure 10 whose equation is

$$N^{0.88} \Delta \epsilon_p = 6.0 \quad (6)$$

All data lying on this line represent specimens that failed without buckling. The data for SCF-tested specimens that failed by buckling lie along the line

$$N^{0.25} \Delta \epsilon_p = 0.24 \quad (7)$$

as determined by using data from the earlier program.¹

The tests on annealed (24 hours at 725°C) Zr-2 indicate an SCF resistance superior to that of cold-worked material on a strain basis of comparison. The TSF-tested annealed material exhibits a resistance superior to cold-worked material, although there is appreciable data scatter in these tests, as seen in Figure 10. Typical microstructures of annealed, SCF-tested Zr-2 are shown in Figure 13a. An identical test on cold-worked material is shown in Figure 13b for purposes of comparison. Both specimens failed by transcrystalline fracture, and both specimens underwent mechanical twinning during fatigue. The fact that twins are evident only near the fracture suggests that a high, and probably complex stress distribution is necessary for their formation. Twinning occurs over a shorter distance in the annealed material than in the cold-worked material. This would be expected if the dislocation density was greater in the cold-worked material (which of course is very logical), causing the high-stress field to be transferred farther from the advancing crack than it would be in the annealed material.

The superiority of the annealed material over the cold-worked material, shown in Figure 10, means that the annealed material has a greater modified ability to withstand plastic strain. This is true even though the annealed material often strain-hardened and the cold-worked material often strain-softened during both TSF and SCF cycling.

If oxidation effects are not present (e.g., when TSF testing in argon), the data in Figure 10 indicate that the fatigue resistance of Zr-2 tested by thermal cycling is equivalent to that of Zr-2 tested by conventional strain cycling. The strong dependence of test results on atmosphere leads to the conclusion that it would be risky to assume that the fatigue resistance of Zr-2 in a liquid-metal or other corrosive atmosphere would be the same as in argon.

FATIGUE TEST RESULTS FOR ZIRCALOY-2

 $\Delta\epsilon_{p,s}$ versus N

FIGURE 10

150×

← Stress →

Neg. #5766

a. TSF-tested at $T_m = 400^\circ\text{C}$, $T_2 = 600^\circ\text{C}$ (Specimen 215-7)

150×

← Stress →

Neg. #5767

b. SCF-tested at 600°C (Specimen 264-10)

COLD-WORKED ZIRCALOY-2
TSF- AND SCF-TESTED IN AIR AT IDENTICAL MAXIMUM TEMPERATURES
Initial wall thickness = 0.020 inch. Fatigue fractures at right.
(For etch see Table V, Appendix D.)

FIGURE 11

150×

← Stress →

Neg. #5731

COLD-WORKED ZIRCALOY-2
TSF-TESTED IN ARGON AT $T_m = 400^\circ\text{C}$, $T_2 = 700^\circ\text{C}$
(Specimen 286-11)

FIGURE 12

150 ×

← Stress →

Neg. #5729

a. Annealed structure (Specimen 297-11)

150 ×

← Stress →

Neg. #5730

b. Cold-worked structure (Specimen 263-10)

ZIRCALOY-2 SCF-TESTED AT 400°C
IN ANNEALED AND COLD-WORKED CONDITIONS
Fatigue fractures shown.

FIGURE 13

D. A387-D (Cr-Mo) Steel

This steel, often designated Croloy 2-1/4, yielded the results shown in Figure 14 and listed in Table VI. The effect of oxidation is evident from the fact that tests in air yielded a failure line represented by

$$N^{0.65} \Delta \epsilon_p = 0.23 \quad , \quad (8)$$

whereas tests in argon yielded data represented by the equation

$$N^{0.58} \Delta \epsilon_p = 0.25 \quad , \quad (9)$$

Failure equation 8 is unaffected by mean temperature of testing and hold time at temperature extremes in the ranges studied. This result is of course desirable, since this steel, when actually applied under conditions that this test program has bracketed but not duplicated, could be analyzed for fatigue life prior to such application.

The results of another TSF research program⁷ on Croloy 2-1/4 are also plotted in Figure 14 for comparison with equivalent data generated in this program. There is relatively poor agreement between the two dashed lines for total strain range ($\Delta \epsilon_t$). The ATL results were obtained using a gage length with a very low temperature gradient ($\sim 10^\circ\text{C}$), whereas the data in reference 7 were obtained using a gage length with a far greater temperature gradient. That is, the low-thermal-gradient tests by ATL allowed measurement of essentially the maximum $\Delta \epsilon_p$ (or $\Delta \epsilon_t$), whereas the high-thermal-gradient tests no doubt measured an average $\Delta \epsilon_t$ that was significantly less than the maximum.

The structures of the TSF- and SCF-tested Croloy 2-1/4 are essentially identical to the structure of the as-received (annealed) material, as is illustrated in Figure 15. The as-received structure is duplicated in the TSF-tested specimen, with the addition of transcrystalline fatigue cracks.

When TSF and SCF test results are plotted on coordinates of cyclic stress range versus N, considerable data scatter is found, as shown in Figure 16. However, the ATL results (solid line) are in rough agreement with data generated elsewhere (dashed lines).⁷

FATIGUE TEST RESULTS FOR AS-RECEIVED A387-D (CROLOY 2-1/4)

 $\Delta\epsilon_{p,s}$ versus N

FIGURE 14

250x

Neg. #5182

a. As-received (annealed) structure

250x

← Stress →

Neg. #5558

b. TSF-tested at $T_m = 400^\circ\text{C}$, $T_2 = 600^\circ\text{C}$
Fatigue fracture at right (Specimen 479)

A387-D STEEL (CROLOY 2-1/4)
TYPICAL STRUCTURES BEFORE AND AFTER FATIGUE TESTING
(For etch see Table V, Appendix D.)

FIGURE 15

FATIGUE TEST RESULTS FOR AS-RECEIVED A387-D (CROLOY 2-1/4)

 $\Delta\sigma_m$ versus N

FIGURE 16

E. A302-B Steel

The data from the TSF and SCF tests on the ferritic pressure-vessel steel, ASTM type A302-B, are plotted in Figure 17 and listed in Table VII. This alloy caused considerable trouble during testing in that unusual data scatter occurred. This scatter could be due to inconsistencies in specimen history, since the raw material was obtained in the form of 4-inch-thick plate, from which slices were flame-cut and forged to bar stock of 1-3/8-inch diameter. Nevertheless, certain conclusions can be drawn from these tests.

The trend of the data definitely indicates that TSF-tested specimens fail sooner than SCF-tested specimens. The effects of hold time and T_m on TSF results are negligible. However, specimens TSF-tested in argon last longer, in general, than their counterparts tested in air. Thus, fatigue life was affected by oxidation, but data scatter prevented observation of the effects of T_m and hold time on oxidation. Also of significance is that the TSF tests in argon roughly correspond to SCF tests in air, indicating that the SCF line represents the true TSF and SCF resistance of the material, excluding oxidation effects.

The structure of A302-B after TSF testing can be significantly different from the as-received structure. For example, the as-received structure shown in Figure 18 is unchanged when the specimen is SCF-tested at 400°C (Figure 19a) or TSF-tested at a maximum T_2 of 700°C (Figure 19b). However, TSF testing at $T_2 = 800^\circ\text{C}$ causes a bainitic-like structure to form (Figure 20a). This formation has been termed bainitic ferrite.⁸ The structure of solution-treated water-quenched A302-B is shown in Figure 20b for comparison. This structure is also probably bainite. The low hardness of the structure is accounted for by the low carbon content of the alloy.

The bainitic structures could have been caused by specimen overheating and not by normal testing. That is, when a crack forms, the region immediately adjacent to the crack is heated less than normal but overheating occurs in the uncracked material. A fatigue crack can thus propagate through an overheated region unless automatic shutdown occurs first. Hence, if specimen overheating were the reason for the appearance of bainitic structures, a nonbainitic structure should be observed in tests free from overheating, as indeed is shown in Figure 21 for a TSF test performed with $T_2 = 800^\circ\text{C}$ and no overheating.

The difference in structure between the surface and the interior illustrated in Figure 21 resembles a gradient in secondary recrystallization. The only plausible reason for such a gradient in grain size is that the small globular grains are cementite. These grains, when located near the surface, lose carbon and transform to ferrite, thus enabling the larger ferritic grains to absorb them by grain growth.

FATIGUE TEST RESULTS FOR AS-RECEIVED A302-B STEEL

 $\Delta \epsilon_{p,s}$ versus N

FIGURE 17

750 ×

Neg. #5102

AS-RECEIVED A302-B STEEL
Hardness: Vickers 26; R_H 19
(For etch see Table V, Appendix D.)

FIGURE 18

250 ×

← Stress →

Neg. #5105

a. SCF-tested at 400°C (Specimen 292)

250 ×

← Stress →

Neg. #4987

b. TSF-tested at $T_m = 500^\circ\text{C}$, $T_2 = 700^\circ\text{C}$ (Specimen 253)

SCF- AND TSF-TESTED A302-B STEEL,
SHOWING FATIGUE FRACTURES BUT OTHERWISE UNCHANGED STRUCTURES

FIGURE 19

150x ← Stress → Neg. #5104

- a. TSF-tested at $T_m = 500^\circ\text{C}$, $T_2 = 800^\circ\text{C}$
Hardness: Vickers 33 (Specimen 251)

100x Neg. #5108

- b. Structure after water quenching from 930°C
Hardness: Vickers 36

A302-B STEEL SHOWING BAINITIC STRUCTURES

FIGURE 20

↑ Stress ↓

1000 ×

Neg. #5021

A302-B STEEL TSF-TESTED AT $T_m = 500^\circ\text{C}$, $T_2 = 800^\circ\text{C}$,
SHOWING ABSENCE OF BAINITIC STRUCTURE
Specimen surface is at right. (Specimen 263)

FIGURE 21

F. Incoloy 800

A very brief test program was performed on a special Incoloy^{*} alloy that has been found to have superior corrosion resistance. This apparently single-phase alloy (see Figure 22) yielded the TSF failure curve shown in Figure 23. The equation of this line,

$$N^{0.44} \Delta \epsilon_p = 0.068 \quad , \quad (10)$$

closely approximates the slope of $\frac{1}{2}$ proposed by Coffin.³ Since many of the failures were due to instability, it is believed possible for the pure-fatigue-failure line to be steeper.

The structure of Incoloy is unaffected by TSF tests, as shown in Figures 24 and 25. However, the fracture path changes from transcrystalline to intercrystalline when T_2 is increased above 650°C.

The total plastic strain energy absorbed is shown in Figure 26 to be proportional to the number of strain reversals (2N) such that the slope is similar in magnitude to that found for common metals tested in conventional fatigue.⁶ While the data points obtained for Incoloy no doubt give a fair approximation of its fatigue resistance, performance of additional tests to confirm the effects of test variables would be desirable.

Data from this brief Incoloy 800 program are summarized in Table VIII.

* Supplied by General Electric Company, San Jose, California.

100 ×

Neg.: #4902

INCOLOY 800 TYPICAL STRUCTURE
(For etch see Table V, Appendix D.)

FIGURE 22

TSF TEST RESULTS FOR AS-RECEIVED INCOLOY 800

 $\Delta\epsilon_{p,s}$ versus N

FIGURE 23

150 ×

← Stress →

Neg. #4937

INCOLOY 800 TSF-TESTED AT $T_m = 500^\circ\text{C}$, $T_2 = 650^\circ\text{C}$

Fatigue fracture shown. (Specimen 6)

FIGURE 24

150 ×

← Stress →

Neg. #4935

INCOLOY 800 TSF-TESTED AT $T_m = 500^\circ\text{C}$, $T_2 = 750^\circ\text{C}$

Fatigue cracks near failure region. (Specimen 5)

FIGURE 25

TSF TEST RESULTS FOR AS-RECEIVED INCOLOY 800
 $N(\Delta W_s)$ versus N

FIGURE 26

G. 403 Stainless Steel

A type 403 martensitic stainless steel was TSF- and SCF-tested in the quenched and tempered (2 hours at 1200°F) condition. The results of these tests are plotted in Figure 27, * from which is seen the excellent agreement between TSF and SCF testing. None of the variables investigated (T_m , hold time, strain rate) affected the failure curve; thus it can be assumed that a designer has a significant amount of leeway in using Figure 27.

The total cyclic strain, $\Delta\epsilon_t$, is plotted in Figure 28. These data also exhibit apparent independence of test conditions, except for the low-temperature SCF tests. The exception is to be expected, as this type of graph should be influenced to some extent by variation in the tendency of the specimen to undergo plastic strain as a result of testing at different temperatures. When Figures 27 and 28 are superimposed, it is seen that beyond 1000 cycles to failure, plastic strain range becomes far smaller than elastic strain range, whereas the reverse is true in tests failing in less than 1000 cycles.

When cyclic maximum stress range is plotted versus N , considerable data scatter is found (see Figure 29). It is obvious, however, that the test conditions at which the material exhibits its greatest strength (its least tendency to undergo plastic strain) are also the conditions at which the material will exhibit the greatest fatigue strength. The data scatter shown in Figure 29 is comparable to the scatter found by other investigators⁴ when plotting $\Delta\sigma$ versus N for TSF tests.

The microstructures of 403 stainless steel are of interest since presumably these structures would be unstable during TSF testing. The as-received (annealed) structure is shown in Figure 30. When this steel is quenched from 1000°C, it hardens to R_C 42-44, and tempering at 649°C (1200°F) brings the hardness down to R_C 20-23. The structure of tempered martensite is identical to that of the TSF-tested ($T_2 = 500^\circ\text{C}$, $T_1 = 100^\circ\text{C}$) specimen shown in Figure 31. Over-tempering for 24 hours at 649°C causes only a nominal further decrease in hardness to R_B 90 and no observable change in structure.

* Results are summarized in Table IX.

TSF testing at progressively higher temperatures changes the structure. For example, Figure 32 shows a specimen tested at $T_2 = 575^\circ\text{C}$, $T_1 = 225^\circ\text{C}$ (zero hold time), with no change in structure. However, when the specimen is cycled between $T_2 = 700^\circ\text{C}$ and $T_1 = 100^\circ\text{C}$, the hardness decreases to $R_B 84$ and the structure becomes more spheroidized, as shown in Figure 33. Hold time does not alter the structure as compared to an identical test without hold time (compare Figures 33 and 34).

The path of fracture is somewhat difficult to observe because of the indistinct structure. It appears, however, that the transcrystalline fracture occurred at temperatures below 600°C . Since tension occurs at low temperatures in TSF testing and compression at high temperatures, a crack has little tendency to propagate at the higher temperatures; hence, there is little tendency for intercrystalline fracture to occur in tests exhibiting stable failure. In tests exhibiting instable failure, on the other hand, there is opportunity for crack propagation at the high temperature of cycling due to distortion, and it is possible that the specimens shown in Figures 33 and 34 ($T_2 = 700^\circ\text{C}$, $T_1 = 100^\circ\text{C}$) had an intercrystalline fracture path.

The physical nature of the TSF test apparatus used in this research was such that thermal strain was the independent variable.* Thus specimens were always in compression at and near T_2 and in tension at and near T_1 (as noted in the preceding paragraph). It was believed that reversing this tension-compression relationship would change the geometric stability of the specimen and cause a different failure line to occur in the instable-failure region on graphs such as Figure 27. Therefore, one SCF test apparatus was modified to make thermal strain the independent variable, so that tension would occur at T_2 and compression at T_1 . For obvious reasons, these tests were called reverse thermal-stress fatigue (RTSF) tests.

The RTSF instable-failure points plotted in Figure 27 do not agree with the TSF instable-failure points because of the greater specimen instability in RTSF. That is, the RTSF compressive stresses at T_1 far exceeded the TSF compressive stresses at T_2 , causing the specimen to buckle after only a few cycles. It is possible to visualize a series of dashed lines similar to those drawn in Figure 27 for various specimen stabilities. The stability of the specimen would be proportional to the ordinate intercept of these dashed lines. The elongation found in tensile tests run at 400°C , when plotted

* Mechanical strain is of course the independent variable in the SCF test apparatus.

at $N = \frac{1}{4}$, lies on the uppermost extrapolated dashed line drawn in Figure 27. Conceivably, the necking down indicative of metallurgical instability represents the least instable condition and therefore lies in a position as predicted above (i. e., above the more instable, instability failures).

In tests that exhibited no distortion after testing (i. e., stable failures), there was no difference in fatigue resistance for TSF and RTSF conditions. This observation has also been made elsewhere.⁹

The total plastic strain energy absorbed during fatigue testing has been found by many investigators to be a function of the path to failure. That is, this quantity cannot be considered a so-called "state variable." The actual variation in the absorbed energy is shown in Figure 35. The relation between total plastic strain energy absorbed, ($N\Delta W$), and one-half total plastic strain ($N\Delta\epsilon_p$), is

$$N\Delta W = M (N\Delta\epsilon_p)^m \quad (11)$$

The quantity $N\Delta W$ can also be theoretically approximated by the relation

$$N\Delta W = 2N \int_0^{\Delta\epsilon_p} \sigma d\epsilon_p \quad (12)$$

Now a common behavior of materials is that they exhibit strain hardening in a tensile test and in any given cycle in a fatigue test, even though a net cyclic softening may occur during fatigue testing. The relation between stress and strain is

$$\sigma = k \epsilon_p^n \quad (13)$$

and by substituting equations 13 and 11 into equation 12, integrating, and rearranging terms, the following equation is obtained:

$$N^{\frac{m-1}{m-n-1}} \Delta\epsilon_p = \left[\frac{2k}{M(n+1)} \right]^{\frac{1}{m-n-1}} \quad (14)$$

Before the values of the exponents and the constants in equation 14 can be obtained, it is necessary to measure n , (the strain-hardening exponent) and k in an actual fatigue test. By taking a hysteresis loop and replotting it on logarithmic coordinates, k and n follow immediately. This method of evaluating k and n is believed superior to determining their values in a tensile test. However, the latter method has been suggested and used successfully by others.^{10, 11} Substituting the average values of k and n from hysteresis loops, and M and m from Figure 35, into equation 14 gives

$$N^{0.47} \Delta \epsilon_p = 5.4 \quad . \quad (15)$$

Since the experimental equation for 403 stainless steel is

$$N^{1.09} \Delta \epsilon_p = 7.2 \quad , \quad (16)$$

the agreement between purely empirical and quasi-theoretical failure equations is seen to be relatively poor. Using values of k and n obtained from 400°C tensile tests did not improve the agreement between theory and experiment.

FATIGUE TEST RESULTS FOR 403 STAINLESS STEEL
 $\Delta\epsilon_{p,s}$ versus N

FIGURE 27

FATIGUE TEST RESULTS FOR 403 STAINLESS STEEL

 $\Delta\epsilon_{t,s}$ versus N

FIGURE 28

FATIGUE TEST RESULTS FOR 403 STAINLESS STEEL
 $\Delta\sigma_m$ versus N

FIGURE 29

250 ×

Neg. #5183

AS-RECEIVED (ANNEALED) 403 STAINLESS STEEL
(For etch see Table V, Appendix D.)

FIGURE 30

100 ×

← Stress →

Neg. #5556

403 STAINLESS STEEL TSF-TESTED AT $T_m = 300^\circ\text{C}$, $T_2 = 500^\circ\text{C}$
Fatigue fracture shown. (Specimen 391)

FIGURE 31

250 x

← Stress →

Neg. #5209

403 STAINLESS STEEL TSF-TESTED AT $T_m = 400^\circ\text{C}$, $T_2 = 570^\circ\text{C}$,
ZERO HOLD TIME AT T_2

Fatigue fracture shown. (Specimen 392)

FIGURE 32

250 ×

← Stress →

Neg. #5193

403 STAINLESS STEEL TSF-TESTED AT $T_m = 400^\circ\text{C}$, $T_2 = 700^\circ\text{C}$,
ZERO HOLD TIME AT T_2 , T_1

Hardness: R_B 84. Fatigue fracture at right. (Specimen 379)

FIGURE 33

250 ×

← Stress →

Neg. #5188

403 STAINLESS STEEL TSF-TESTED AT $T_m = 400^\circ\text{C}$, $T_2 = 700^\circ\text{C}$,
60-SEC HOLD TIME AT T_2 , T_1

Hardness: R_B 84. Fatigue fracture at right. (Specimen 353)

FIGURE 34

H. 304 and 304-L Stainless Steels

The results of this program were discussed in detail in an earlier report,² but a brief discussion of them is in order because the testing of 304 and 304-L austenitic stainless steel occupied 40 to 50% of the program. The total data from these tests on 304 and 304-L are plotted in Figures 36 and 37, respectively, and summarized in Tables X and XI, respectively.

An appreciable effort was expended in obtaining values for apparent fatigue resistance as a function of measurement gage length, ℓ_o' . This effort was rewarded by the development of the strain-measuring device and technique discussed earlier in this report. The difference in apparent fatigue life is clearly shown for 304-L stainless steel in Figure 38, where the solid points represent the true (short ℓ_o') fatigue resistance and the open points the apparent (long ℓ_o') fatigue resistance.

The superiority of annealed material over cold-worked (18% reduction in area) material is particularly evident in Figure 38 on a strain basis of comparison. On a stress basis of comparison, the cold-worked material is superior (see Figure 39). These trends are identical to those found for Zircaloy-2. In fact, the similarity between Zr-2 and 304-L stainless steel goes even further. For example, both alloys have a tendency to seek a characteristic hardness during cycling. In Figure 40, it is seen that cold-worked material fatigue-softened at low mean temperature (and relatively high $\Delta\epsilon_p$), whereas annealed material hardened under identical conditions. It has been noticed that the tendency to harden or soften is also a function of the strain amplitude and that at low strain amplitudes cold-worked 304-L will harden, whereas at high strain amplitudes it will soften. This behavior was also exhibited by Zr-2¹ and has been observed in mild steel¹² in conventional fatigue.

The difference in the structures of TSF-tested, annealed and cold-worked 304-L is illustrated in Figure 41. The grain-size difference is obvious, as is the tendency for transcrystalline fracture path. Increasing the mean temperature of testing had no effect on the structures, as seen in Figure 42.

The wide grain boundaries exhibited by both annealed and cold-worked TSF-tested 304 and 304-L are not typical of, for example, 304-L in either the annealed or the as-received (cold-worked) condition before TSF testing. The latter two structures are

shown in Figure 43 for comparison with Figures 41 and 42. It was believed that plastic strain during TSF testing decreased the relative misorientation between grains, thus widening the grain boundaries. This hypothesis was investigated further by X-ray diffraction analysis. A (111) pole figure for as-received (cold-worked) material did not differ from a similar pole figure for a cold-worked specimen that had undergone 10 in./in. total plastic strain during TSF testing. However, the relatively large grain sizes (30 to 40 microns) of these specimens no doubt prevented the true texture from being observed.

The main cause of apparent grain-boundary widening is precipitation at grain boundaries, and it is certain that discontinuous carbide precipitation occurred in most TSF tests in 304 and 304-L steels. A typical region of precipitation is pictured in Figure 44.

The significant difference in grain size between the cold-worked and the annealed structures was not considered to have influenced fatigue resistance, in view of the results on annealed (2 hours at 2100°F) large-grain and annealed (2 hours at 1900°F) small-grain 304-L. The study showed no difference in test results over a failure range of 10^1 to 10^4 cycles.

The relative independence of mean temperature of testing displayed by the 403 stainless steel test results was duplicated for the cold-worked 304-L stainless steel. However, the annealed, small-grain 304-L did exhibit an unmistakable T_m effect, as seen in Figure 45. Metallographic examination of tested specimens did not reveal any difference in structures at different mean temperatures. The high T_m tests, failing sooner than the low T_m tests, must somehow reduce the ability of the steel to absorb plastic strain. This could come about because of a change in the distribution of plastic strain. That is, at the higher T_m more strain could be taking place in the grain boundaries than occurred there at a lower T_m , and if the grain boundaries had a low ability to absorb this strain, fracture would occur through the boundary. This explanation requires that in the low N region of failure, where even the low T_m tests have a relatively high T_2 , the low and high T_m test results approach each other, as indeed they do in Figure 45. Another reason why the low and high T_m test results approach each other in the low N region is that TSF specimens failing in this region

exhibit instable (buckling) failures. Such failures are probably more insensitive to the plastic-strain-absorbing properties of a material than are stable failures. In any case, the fractures in the low N region are always intercrystalline, as shown for example in Figure 46.

Since the cold-worked material does not show a T_m effect and the annealed material does, it is difficult to make the above reasoning for the T_m effect compatible with the absence of such an effect. That is, it is not known why, in the cold-worked material, plastic strain does not tend to shift from within the grain to the grain boundaries with increasing T_m . To resolve this anomaly, it would be necessary to make an electron-microscope study of tested specimens.

The effect of a hold time at temperature extremes, and in particular at T_2 , was found to be inconsequential for 304-L steel, as shown in Figure 47 for cold-worked material. This result is to be expected on a strain basis of comparison where the differences in structural changes (i. e., dislocation arrangements, crack propagation, etc.) occurring between hold and no-hold tests are in degree, not kind. One could expect a hold-time effect on a strain basis of comparison if an incubation (or nucleation) time existed for the phenomenon occurring.

During the course of research on 304-L, an effort was made to ascertain the effects of specimen size (in this case, specimen wall thickness). Since the fatigue cracks usually start at free surfaces, it is logical to expect an increasing fatigue resistance with a decreasing ratio of surface area to volume.* However, no effect could be found for 304-L in either the pure-fatigue-failure region or, surprisingly, the instable-failure region (see Figure 48). This lack of effect could be due to the relatively small change in volume employed; SA/V was varied from only 50 to 100 inch^{-1} . It could also be due to a canceling of SA/V effect by a pure volume effect of the type¹³

$$\frac{N_1}{N_2} = \left[\frac{V_2}{V_1} \right]^{1/\eta} \quad (17)$$

* This ratio, SA/V, is inversely proportional to specimen wall thickness, t' , and equal to $2/t'$.

It was stated earlier that carbide precipitation (sensitization) occurred at the 304-L grain boundaries. Thus, it was considered possible to observe a stress-corrosion effect. To test this hypothesis, tests were run in air and in vacuum (10^{-5} torr). Tests were also run in argon to minimize oxidation caused by air leaking into the test fixtures. Regardless of the atmosphere, or lack thereof, the TSF test results were the same. Hence, neither stress corrosion nor surface oxidation were of significant magnitude to affect test results.

A number of SCF tests were run on 304-L stainless steel for comparison with TSF results. A plot of the data points is presented in Figure 48. Only on the cold-worked material do TSF and SCF agree. Typical structures of SCF- and TSF-tested, cold-worked 304-L are shown in Figure 49. In both specimens, fracture was transcrystalline and some grain-boundary widening due to carbide precipitation occurred as a result of testing.

A few subzero TSF tests were run on 304-L. In these tests, a T_1 temperature of -50°C was obtained by cooling with liquid CO_2 . Data from these tests conform to data obtained at higher mean test temperatures.

Plastic-strain-energy data were also obtained on 304-L stainless steel. The data for the cold-worked material are shown in Figure 50. Using the slope and intercept values from the graph, together with values of k and n obtained in tensile tests, equation 14 becomes

$$N^{2.58} \Delta\epsilon_p = 0.97 \quad , \quad (18)$$

as compared to the experimentally determined equation from Figure 38,

$$N^{1.29} \Delta\epsilon_p = 43 \quad . \quad (19)$$

Clearly, there is no agreement between these two equations. In fact, all attempts to derive quasi-theoretical failure equations were unsuccessful, insofar as obtaining agreement with equation 19. For example, if the relation

$$N\Delta W = M (\Delta\epsilon_p)^m \quad (20)$$

is plotted and values for k and n are taken from stable hysteresis loops, equation 14 becomes

$$N^{0.25} \Delta \epsilon_p = 0.026 \quad . \quad (21)$$

The rather significant discrepancies among the equations might be interpreted as being caused by poor or incorrect data. However, it is believed that this is not the case, for when data from other sources are available, they compare reasonably well with the data generated in this program. In Figure 36, the TSF results on 304 stainless steel are seen to closely approximate results from another laboratory.

FATIGUE TEST RESULTS FOR 304 STAINLESS STEEL

 $\Delta\epsilon_p$ versus N

FIGURE 36

FATIGUE TEST RESULTS FOR 304-L STAINLESS STEEL,
COMPARING LONG AND SHORT MEASUREMENT GAGE LENGTHS
AND ANNEALED AND COLD-WORKED CONDITIONS
 $\Delta\epsilon_p$ versus N

FIGURE 38

FATIGUE TEST RESULTS FOR 304-L STAINLESS STEEL,
COMPARING ANNEALED AND COLD-WORKED CONDITIONS
 $\Delta\sigma_m$ versus N

FIGURE 39

TSF TEST RESULTS FOR 304-L STAINLESS STEEL
 $\Delta\sigma_m$ versus Cycles

FIGURE 40

100 × ← Stress → Neg. #5048

a. Annealed structure (2 hours at 1038°C) (Specimen 54)

100 × ← Stress → Neg. #4628

b. Cold-worked structure (Specimen 86)

304-L STAINLESS STEEL IN ANNEALED AND COLD-WORKED CONDITIONS
TSF-TESTED AT $T_m = 400^\circ\text{C}$, $T_2 = 600^\circ\text{C}$

Fatigue fractures at right.
(For etch see Table V, Appendix D.)

FIGURE 41

100 ×

← Stress →

Neg. #5049

a. Annealed structure (2 hours at 1038°C) (Specimen 39)

100 ×

← Stress →

Neg. #5057

b. Cold-worked structure (Specimen 80)

304-L STAINLESS STEEL IN ANNEALED AND COLD-WORKED CONDITIONS
 TSF-TESTED AT $T_m = 600^\circ\text{C}$, $T_2 = 800^\circ\text{C}$

FIGURE 42

100 ×

Neg. #4377

a. Annealed structure (2 hours at 1038°C)

100 ×

Neg. #4362

b. As-received (cold-worked) structure

304-L STAINLESS STEEL IN ANNEALED AND COLD-WORKED CONDITIONS
BEFORE TSF TESTING

FIGURE 43

500×

Neg. #5064

TSF-TESTED 304 STAINLESS STEEL,
SHOWING CARBIDE PRECIPITATION AT GRAIN BOUNDARIES
(Specimen 5)

FIGURE 44

TSF TEST RESULTS FOR ANNEALED, SMALL-GRAIN 304-L STAINLESS STEEL,
SHOWING DEPENDENCE ON MEAN TEMPERATURE OF TESTING

$\Delta\epsilon_p$ versus N

FIGURE 45

100×

← Stress →

Neg. #4643

304-L STAINLESS STEEL TSF-TESTED AT $T_m = 600^\circ\text{C}$, $T_2 = 850^\circ\text{C}$,
SHOWING INTERCRYSTALLINE FRACTURES
(Specimen 57)

FIGURE 46

TSF TEST RESULTS FOR COLD-WORKED 304-L STAINLESS STEEL,
SHOWING INDEPENDENCE OF HOLD TIME
 $\Delta\epsilon_p$ versus N

FIGURE 47

FATIGUE TEST RESULTS FOR 304-L STAINLESS STEEL, SHOWING INDEPENDENCE OF SPECIMEN WALL THICKNESS AND COMPARING TSF WITH SCF ($\Delta\epsilon$ versus N)

FIGURE 48

100x

← Stress →

Neg. #5126

- a. SCF-tested at 500°C. Area typical of region undergoing tensile strain after fatigue failure is essentially complete in an adjacent region. (Specimen 124)

100x

← Stress →

Neg. #5133

- b. TSF-tested at $T_m = 400^\circ\text{C}$, $T_2 = 550^\circ\text{C}$ (Specimen 109)

COLD-WORKED 304-L STAINLESS STEEL,
COMPARING EFFECTS OF SCF AND TSF TESTING
Fatigue fractures shown.

FIGURE 49

TSF TEST RESULTS FOR COLD-WORKED 304-L STAINLESS STEEL
 $N(\Delta W)$ versus $N(\Delta\epsilon_p)$

FIGURE 50

GENERAL DISCUSSION OF RESULTS

The general trend in fatigue-resistance equations is for a slope of approximately $\frac{1}{2}$ to 1, as seen in Figure 51. While in certain cases there is little difference in fatigue resistance among alloys, it has been shown that an optimum condition often exists for fatigue resistance in any given alloy. This optimum condition depends on the basis of comparison and usually reverses with a change in this basis.

Often, conflicting trends in the optimum condition for fatigue resistance have been found on the same basis of comparison. This latter occurrence is graphically illustrated in Figure 52 for 304-L and a Nimonic alloy.¹⁴ On this graph, the fatigue resistance of the low-strength (high- T_m) Nimonic alloy is superior to the resistance of the high-strength (low- T_m) alloy in the low-N range where $\Delta\epsilon_p$ approaches $\Delta\epsilon_t$. The 304-L test results are seen to conflict with the Nimonic results in this region. Conversely, in the high-N range, the TSF resistance of both 304-L and Nimonic in the high-strength state is superior, presumably because in this range $\Delta\epsilon_e$ approaches $\Delta\epsilon_t$. Therefore, the material condition with the greatest strength (lowest tendency to undergo plastic strain) should have the best resistance.¹⁴

It is interesting to note that the Nimonic data indicate that the ability to absorb plastic strain is proportional to the $\sqrt{N}$ in certain cases and to $0.2 \sqrt{N}$ in other cases. In other words, the ability to absorb plastic strain increases with increasing N. Consequently, if the materials in the high-strength condition were superior in TSF resistance, they would appear to be so by virtue of having a greater ability to absorb plastic strain, not by virtue of having a lower tendency to undergo plastic strain. The apparent requirement of a greater ability to absorb plastic strain (for a greater TSF resistance) in the seemingly least ductile (high-strength, low- T_m) material condition is rather difficult to reconcile for the Nimonic alloy. However, tensile-test data on annealed 304-L indicate a greater ductility at 400°C than at 600°C, thus substantiating the trend shown in Figure 52.

The fallacy in ascribing superior TSF resistance to low- T_m tests (in the high-N range) to the greater material strength at low T_m is evident when one considers strain as the independent variable. Hence, if a material must undergo a given $\Delta\epsilon_t$, the high-strength material will obviously be under a far greater cyclic stress than the low-strength material and, other conditions being equal, could fail more quickly.

It would no doubt be possible to predict a dynamic dislocation model to explain the change in ability to absorb plastic strain. Such a model would be of academic interest only, since no proof would be available and since dislocation theory can invariably explain any conceivable set of experimental observations.¹⁵ Clearly, an electron microscope would have been invaluable in studying dislocation structures in a test program where strain (in this case, thermal strain) was the independent variable. Unfortunately, access to an electron microscope was very limited. The few specimens whose structures were observed by transmission microscopy confirmed the often noted conclusion¹⁶ that specimen heterogeneity over small distances (10 to 100 microns) requires almost a statistical study of specimens. One other benefit obtained was the development of an experimental technique of making TSF-tested specimens into electron-microscope specimens.¹⁷ Further electron-microscope study of the TSF and SCF specimens tested in this program is strongly recommended.

There is often a tendency to attempt to relate tensile-test data and TSF or SCF data.¹¹ The major differences in tensile tests and fatigue tests are that strain is cyclic (alternating in direction) in the latter test but not in the former, and that the tensile test is usually of very short duration in comparison to the fatigue test. The TSF failure line, extrapolated to $N = \frac{1}{4}$, could accurately predict the tensile-test fracture ductility if alternating plastic strain and unidirectional plastic strain caused identical crack initiation and propagation characteristics. Since tensile tests usually exhibit necking down (i. e., instability), it would appear logical for the fatigue test also to exhibit necking down if these two types of tests were to agree with each other. However,* no necking down due to fatigue strain has been observed. Hence, it is felt intuitively that agreement between TSF and tensile-test data would be fortuitous. However, it is also believed that by proper control of the tensile-test condition it would be possible for tensile-test ductility to lie on the extrapolated TSF $\Delta\epsilon_p$ versus N curve at $N = \frac{1}{4}$. Tensile tests were performed during this program, and as expected, they did not agree with TSF data (see Figures 27 and 48). Whether or not the ability to absorb plastic

* The slight necking down in Figure 49b is not due to fatigue strain, but rather to tensile failure after fatigue failure.

strain is of paramount importance in determining the ductility of a specimen in a tensile test is unclear, because of the occurrence of necking down. Conversely, in a TSF or SCF test, the ability to absorb cyclic plastic strain is of prime importance in determining specimen life.

The TSF resistance from all known (usable) research are plotted in Figure 53. The actual test conditions used to obtain these data vary greatly, as in fact do the definitions of failure, and Figure 53 is not intended to infer the superiority of a particular material over another material. The reader wishing to choose an alloy for TSF resistance should consult the original references.

Figure 53 provides some interesting observations on the general form of TSF failure equations. For example, the slopes of lines vary from ~ 0.2 to ~ 1.2 with no slope being dominant. The predicted fracture ductilities ($\Delta\epsilon_p$ at $N = \frac{1}{4}$) vary immensely. However, it has already been stated that tensile-test data do not (and probably should not) lie on the extrapolated TSF lines. Whether or not there should be a universal slope for these lines is open to debate. The derived failure equation (14) requires the exponent to be related to the constant (the right side of the equation); hence there could be no universal slope. It would, of course, be possible to derive an equation of the form

$$N^{k'} \Delta\epsilon_p = \text{constant} \quad , \quad (22)$$

where k' would be independent of the constant.

In connection with the question of a proper slope for the failure equation, in the 304-L program it was noted that the effect of cold work was to change the exponent far more than the constant, whereas T_m had more effect on the constant than on the exponent. It therefore appears logical to conclude that there is no universal value for the exponent.

The usefulness of the plastic strain graphs presented herein is dependent on the ability of a designer to estimate his expected cyclic plastic strain. This is clearly not an unrealistic demand, since the usefulness of conventional-fatigue-data plots ($\Delta\sigma$ versus N) depends on the ability of the designer to calculate his estimated cyclic stress, and if stress is known, strain is also usually known. However, since most fatigue test results in existence are on graphs of stress amplitude or range versus cycles to failure, it is propitious to summarize the test results obtained in this program in like manner.

Thus, in Figure 54, data from $T_m = 400^\circ\text{C}$, zero-hold-time TSF tests are plotted on a stress basis. A change in T_m would greatly alter the position of the lines. Therefore, unless a designer has an application involving cycling stress around 400°C mean temperature, he could not use Figure 54.

Figures 54 and 51 together provide interesting conclusions on the tendency to undergo plastic strain and the ability to withstand plastic strain, respectively. It is seen that Zircaloy-2 is inferior to the other materials on both bases of comparison, whereas 304-L (cold-worked) is generally superior to the other alloys on both bases of comparison. This latter observation implies that in a comparison among different alloys, the alloy with the greatest toughness^{*} will exhibit optimum fatigue properties. To test this hypothesis, a graph of toughness, as measured by $(\Delta\sigma_m)(\Delta\epsilon_p)$ versus N , is shown in Figure 55. This graph generally follows the pattern of results indicated in Figure 51, with a few exceptions; for example, the reversal of superiority (as a function of N) between Incoloy 800 and A387-D steel (Croloy 2-1/4). The necessity to fix N prior to determining the best alloy on a toughness or strain basis is made obvious in Figures 55 and 51. It should be emphasized in connection with the above graphs that in choosing an alloy solely on a fatigue-resistance basis, either the stress or the strain graph (Figure 54 or Figure 51, respectively) would be used, relegating a toughness basis of comparison merely to academic interest.

The propagation of a fatigue crack appears to be a rather rapid phenomenon, occurring late in the life of a specimen. This observation is supported by the continuous-load recordings made on all (both TSF and SCF) fatigue tests in this program. These recordings indicated a constant cyclic load, followed by a rapid decrease in cyclic tensile load coincident with the appearance of a crack in the specimen. Such a form of fatigue fracture is more indicative of low-cycle fatigue (high-strain-amplitude fatigue) than high-cycle fatigue.¹⁸

* As determined, for example, by measuring the area under the tensile-test stress-versus-strain curve.

SUMMARY OF TSF DATA FOR ALLOYS TESTED IN THIS PROGRAM
 $\Delta\epsilon_{p,s}$ versus N

FIGURE 51

TSF TEST DATA FOR ANNEALED 304-L STAINLESS STEEL AND NIMONIC
($\Delta\epsilon_t$ versus N)

FIGURE 52

SUMMARY OF ALL PERTINENT TSF DATA IN TECHNICAL LITERATURE
 $\Delta\epsilon_p$ versus N

FIGURE 53

SUMMARY OF TSF DATA FOR ALLOYS TESTED IN THIS PROGRAM

$\Delta\sigma_m$ versus N

FIGURE 54

ESTIMATE OF TOUGHNESS AS A FUNCTION OF FATIGUE LIFE

$\Delta\sigma_m (\Delta\epsilon_{p,s})$ versus N

FIGURE 55

CONCLUSIONS

A. On Particular Alloys

1. The TSF and SCF test results on Zircaloy-2 agree substantially after oxidation effects are removed. The fatigue-failure line is relatively independent of the test temperature and hold time.

2. The mechanical twinning of Zircaloy-2 is one of the few evidences of plastic strain obtained metallographically in this program. The degree of localization of twins at the fracture region is apparently a function of the initial dislocation density in the alloy. The localization is more pronounced in annealed material, presumably due to the dissipation of high stresses by dislocation creation and movement.

3. The TSF results on A387-D steel (Croloy 2-1/4) are unaffected by the mean temperature of testing and the hold time at temperature extremes. Thus, applications involving conditions that are bracketed by, but not duplicated by, the conditions covered in this test program could utilize the data generated with confidence.

4. A302-B steel specimens subjected to SCF tests did not fail as early as specimens subjected to TSF tests, and those tested in argon did not fail as early as those tested in air. Both of these observations are indicative of an oxidation effect. In TSF tests, however, neither hold time nor T_m had any discernible effect, indicating the absence of an oxidation effect. It must be concluded that data scatter precludes accurate separation of the effects of test variables on this alloy.

5. The brief survey of a special Incoloy alloy (800) yielded a TSF failure equation approximating the 'classical' prediction that slope = $\frac{1}{2}$. However, it is believed that specimen buckling shortened the low-cycle fatigue life and that the pure-fatigue-failure (no-buckling) equation would have a steeper slope.

6. TSF and SCF tests on 403 stainless steel show remarkable agreement. The fatigue-failure line is quite independent of T_m , T_2 , hold time, and strain rate within the range of the variables studied.

7. Reversing the direction of stressing such that tension occurs at T_2 instead of T_1 does not affect the TSF resistance of 403 stainless steel in the stable-failure mode.

8. It was not possible to get agreement between the empirical failure equation and the quasi-theoretical failure equation for 403 stainless steel.

9. Annealed 304-L stainless steel proved to be superior to cold-worked 304-L on a strain basis of comparison. The reverse is true on a stress basis of comparison.

10. In general, neither T_m nor hold time has any effect on plots of $\Delta\epsilon_p$ versus N for 304-L. These results are indicative of the occurrence of phenomena with essentially no incubation time. Therefore, structural changes resulting from cyclic strain are the same (except for degree of advancement) in all tests. However, T_m does have an effect on low-temperature-annealed 304-L. This latter result can be explained by assuming that increased strain occurs in grain boundaries at the high T_m 's and that grain boundaries have a low ability to withstand plastic strain.

11. An increase in the annealing temperature of 304-L, resulting in different grain sizes, does not produce an observable effect on TSF life. Since increasing the annealing temperature not only increases the grain size, but also decreases the average grain-boundary energy, it is not possible to state with certainty that grain size per se will not affect TSF results.

12. Changes in the surface-area-to-volume ratio has no observable effect on TSF life. However, a volume effect could have canceled the surface-area-to-volume effect.

13. Agreement in results between TSF and SCF tests on 304-L was obtained only for the cold-worked material. It is believed that further study of annealed material would also indicate agreement.

14. It was not possible to get agreement between the empirical failure equation and the quasi-theoretical failure equation for 304-L (cold-worked).

B. General

1. The equivalence between TSF and SCF test results is to be expected on graphs of $\Delta\epsilon_p$ versus N whenever identical phenomena (microscopic) occur within the specimen during TSF and SCF testing.

2. It is generally found that the ability to withstand plastic strain ($2N\Delta\epsilon_p$) is proportional to the number of cycles to failure to some positive power between 1 and 0.1. Hence, if strain is the independent variable in a particular application, the superior material, or material condition, is the one with the greatest ability to withstand cyclic plastic strain and not necessarily the one with the least tendency to undergo plastic strain.

3. The material, or material condition, possessing the least tendency to undergo plastic strain would be superior in an application where cyclic stress is the independent variable, other things being equal. In essence, this conclusion means that for a given cyclic stress, the material with strength sufficient to make strain all elastic would be superior to a material with lesser strength that allowed some cyclic plastic strain to occur. On the other hand, if plastic strain does occur in both high- and low-strength materials cycling through a fixed stress range, it does not immediately follow that the high-strength material will have a greater fatigue life since the ability to withstand plastic strain again becomes a deciding factor. Thus, even though the apparent relative fatigue resistance changes when graphs of $\Delta\epsilon_p$ versus N are compared with graphs of $\Delta\sigma_m$ versus N for TSF tests that are independent of thermal strain (see, for example, conclusion 9 on particular alloys), if stress were made the independent variable, it is quite possible that the change in superiority would not occur.

4. The slopes of the fatigue-failure equations are apparently not a universal constant of $\frac{1}{2}$ but are variable within the approximate limits of 0.2 to 1.2.

5. The method of fatigue failure appears to be a rapid propagation of a crack such that the specimen remains essentially intact for the majority of its fatigue life. This method is considered operative only in the high $\Delta\epsilon_t$ range, that is, in tests lasting relatively few cycles (under 10^6), as did all tests performed in this study.

6. The relatively low scatter of fatigue data (in comparison to high- N fatigue, $>10^6$ cycles) found in this program is indicative of low- N fatigue. This small amount of scatter is, of course, ideal for ascertaining the effects of test variables on TSF or SCF life.

7. In high- N fatigue, it becomes essentially impossible to measure cyclic plastic strain. Since cyclic total strain does not give an indication of damage, it is desirable to rely on cyclic stress measurements for representing fatigue life. Conversely in low- N fatigue, cyclic plastic strain is the most accurate means for measuring fatigue damage and for representing general fatigue life.

8. The superior fatigue resistance of a particular alloy condition exists over all failure (N) ranges. In none of the alloys studied in this program did an inferior material condition become superior after a change in the failure range. However, alloys exhibiting superiority of fatigue resistance over certain other alloys in one failure range often became inferior to these same alloys in a different failure range. This conclusion is not considered to be universal to all alloys or alloy conditions.

RECOMMENDATIONS FOR FUTURE STUDIES

The considerable amount of time spent in this program on the effects of changing test conditions on TSF life has elucidated the complexity of the TSF problem. Much emphasis has been placed on the ability to absorb plastic strain as a prime factor in determining fatigue life. Accordingly, it is recommended that further studies be performed on TSF in which strain and temperature can be varied independently, and in which both tensile and bending strains are operative.

The complex interaction of test and material conditions on the ability to absorb plastic strain makes it highly desirable to observe the distribution of plastic strain within the specimen. Thus the application of transmission electron microscopy to TSF-tested specimens would be of inestimable value, and it is strongly recommended that future studies apply such a method.

APPENDIX A

LIST OF SYMBOLS

C	= material constant (related to ductility, $N^{k'} \Delta \epsilon_p = C$) (dimensionless)
E	= modulus of elasticity (psi)
f	= pure-fatigue failure
I	= intercrystalline fracture
i	= instability failure (where not used as subscript)
k	= material constant ($\sigma = k \epsilon_p^n$)
k'	= material constant ($N^{k'} \Delta \epsilon_p = C$)
ℓ_o	= measurement gage length (length over which strain readings are taken) (in.)
M	= proportionality constant = $\frac{N(\Delta W)}{(N \Delta \epsilon_p)^m}$ (in. - lb/in. ³)
m	= slope of $N(\Delta W)$ versus $N(\Delta \epsilon_p)$ curve
N	= cycles to failure
n	= strain-hardening exponent
NF	= no failure
P	= load (lb)
RTSF	= reverse thermal-stress fatigue (i.e., tension occurring at T_2 , compression at T_1)
SA	= specimen surface area
SCF	= strain-cycling fatigue (conventional fatigue)
T	= temperature
TC	= transcrystalline fracture
T_m	= mean temperature of cycling = $(T_2 + T_1)/2$
T_1	= lowest temperature of cycling
T_2	= highest temperature of cycling
T_3	= temperature at which zero stress occurs on heating
T_4	= temperature at which zero stress occurs on cooling
ΔT	= temperature change = $T_2 - T_1$
TSF	= thermal-stress fatigue

t' = specimen wall thickness (in.)
 V = specimen volume
 ΔW = plastic strain energy per cycle (from area of stress-strain hysteresis loop)
(in. -lb/in. ³)
Subscript l - obtained from long measurement gage length
Subscript s - obtained from short measurement gage length
 Y = measured width of hysteresis loop at zero load
 Z = measured unconstrained ℓ_o' movement between T_3 and T_4
 α = coefficient of thermal expansion (averaged over $T_2 - T_1$) (in. /in. -°C)
 ϵ_p = plastic strain (in. /in.)
 $\Delta \epsilon$ = strain range (in. /in.)

<u>1st subscript</u>	<u>2nd subscript</u>
e = elastic	l = long measurement gage length
p = plastic	s = short measurement gage length
t = total	i = measurement during any given cycle

 η = material constant relating the number of large flaws to the number of small flaws
 σ = stress (psi)
 $\Delta \sigma$ = stress range (tension plus compression) per cycle at half life (psi)
 $\Delta \sigma_m$ = maximum stress range (tension plus compression) per cycle at half life (psi)
 $\Delta \sigma_r$ = minimum stress range (tension plus compression) per cycle at temperature extremes - measured at half life (psi)
 ψ, ϕ, ξ = function signs

APPENDIX B

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APPENDIX C
REPORTS ISSUED DURING PROGRAM

Quarterly Technical Progress Report, January-March 1962, ATL-A-130, EURAEC 336
Quarterly Technical Progress Report, April-June 1962, ATL-A-131, EURAEC 335
Quarterly Technical Progress Report, July-September 1962, ATL-A-133, EURAEC 481
Quarterly Technical Progress Report, October-December 1962, ATL-A-134, EURAEC 555
Quarterly Technical Progress Report, January-March 1963, ATL-A-135, EURAEC 632
Topical Report, 10 May 1962, * ATL-A-136, EURAEC 725.
Quarterly Technical Progress Report, April-June 1963, ATL-A-137, EURAEC 756
Quarterly Technical Progress Report, July-September 1963, ATL-A-139, EURAEC 841
Quarterly Technical Progress Report, October-December 1963, ATL-A-140, EURAEC 946
Quarterly Technical Progress Report, January-March 1964, ATL-A-142, EURAEC 1030
Topical Report, 10 April 1964, ATL-A-141, EURAEC 1034
Quarterly Technical Progress Report, April-June 1964, ATL-A-143, EURAEC 1135

* Report should have been dated 1963.

APPENDIX D
TABLES

TABLE I
COMPOSITION OF TEST PROGRAM ALLOYS
(wt. % except as noted)

Alloy	As-received Condition	C	Mn	Si	P	S	Cr	Ni	Cu	Mo	Co	Fe	Sn	Hf	N ₂	O ₂	Zr
304-L Stainless Steel	Cold-worked (18% RA) [†]	0.018	1.32	0.65	0.009	0.011	18.22	10.51				Bal.					
304 Stainless Steel	Annealed	0.05	0.88	0.50	0.028	0.011	18.71	9.84	0.30	0.25		Bal.					
403 Stainless Steel	Annealed	0.10	0.51	0.30	0.008	0.009	12.47	0.34				Bal.					
A302-B Steel	Normalized	0.22	1.3	0.20	0.040	0.040				0.5		Bal.					
A387-D Steel (Croloy 2-1/4)	Annealed	0.10	0.48	0.36	0.014	0.017	2.24			0.94		Bal.					
Incoloy 800	Annealed	0.06	1.08	0.64	0.004	0.006	20.9	32.9	0.02		0.025	Bal.					
Zircaloy-2	Cold-worked (27% RA) [†]	0.024 [*]		0.003 [*]			0.07 [*]	0.06 [*]				0.128 [*]	1.41	0.0045 [*]	0.0055 [*]	0.12 [*]	Bal.

* Atomic percent.

† RA = reduction in area.

TABLE II
TESTS ON 304-L STAINLESS STEEL FOR VCS AND CONVENTIONAL SPECIMENS

Spec. No.	Test Conditions	T ₂ (°C)	T ₁ (°C)	T _m (°C)	ΔT (°C)	Δσ _m (psi)	Δσ _r (psi)	N	Strain Range, Δε (in./in. -cycle)				Failure Type †	Fracture Path ‡	ΔW _s (in.-lb/in. ³)
									Δε _{t,l} [*]	Δε _{t,s} [*]	Δε _{p,l} [*]	Δε _{p,s} [*]			
VCS Specimens															
144X	I. A. 3. a. <u>1</u> . a)-1b							6,900	.0032	.0052	.0027	.00044	f		
149X	-1b												f		
141X	-1c					97,000	97,000	3,655	.0043	.00546		.0023	f		
140X	-1d					118,900	118,900		.0075	.00706	.0030	.0039	i		
142X	-1d					115,000	115,000	165	.0073	.00466	.0029	.0028	i		
148X	-1d					116,700	116,700	100	.0088	.0078	.0029	.005	i		
146X	-3a					41,000	41,000	76,600	.0024	.0015	.00021	.00024	f		
150X				600	300	63,500	63,500	3,550	.0037	.00364	.00046	.00088	f		
113X	-3b					75,600	75,600	212	.0049		.0015	.0021	f	TC	
143X	-3b					77,180	77,180	170	.0056	.0061	.0021	.0024	i		
145X	-3c					70,260	65,500	46	.0074	.0227	.0055	.0204	i		
151X	-3c												i		
147X	-3d					72,350	63,000	20		.0214		.0214	i		
114X	I. A. 3. a. <u>1</u> . c)-3b			600	400	64,200	60,000	25	.0058		.0037		f		
Conventional Specimens															
212	I. A. 3. a. <u>1</u> . a)-	650	-50	300	700	124,900	123,300	90	.0112	.0109	.0056	.0045	i	TC, IC	
209	"	450	-50	200	500	>3,800	110,000								
211	I. A. 3. a. <u>1</u> . c)-	740	460	600	280			900	.0044	.00903	.0017	.0055	f, i		
216	I. A. 3. a. <u>3</u> . c)-1cc	650	150	400	500	88,700	84,460	1,915	.0058	.0074	.0013	.0035	f	TC	
217	"	675	125	400	550	93,200	87,000	720	.0068	.0114	.0015	.0063	i	TC, IC	

* Obtained from diametral measurements for VCS specimens.

† f = pure fatigue; i = instability.

‡ TC = transcrystalline; IC = intercrystalline.

TABLE III
PROGRAM TEST CONDITIONS

EXAMPLE: I.A.1.a.1.a)-1a indicates austenitic stainless steel, type 304-L, annealed (small grain), tested in air, 20-mil wall thickness, thermal-stress-fatigued with zero hold time at temperature extremes at a strain rate of 0.0005 in./in.-sec, temperature limits of cycling 500-300°C.

NOTE: ALL TEMPERATURES IN °C.

TABLE IV
(page 1 of 2)

TESTS ON ZIRCALLOY-2

Spec No.	Test Conditions			$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)		Failure Type *
		T_2 (°C)	T_1 (°C)				$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$	
258-9	V. A. 3. a. 1. c†)	600	400	14,560	14,560	4,170	.00106	.000021	f
281-10		600	200	24,900	24,900	1,175	.00173	.00008	f
215-7		600	100	27,200	26,500	2,020	.00219	.000050	f
259-9		600	100	31,600	30,900	510	.00316	.00030	f
283-10	V. A. 3. a. 1. a)	600	200	23,400	23,400	5,414	.00167	.000057	f
229-9		600	100	27,800	27,800	3,687	.00205	.00011	f
211-7		500	100	23,550	23,550	29,860	.00130	.000035	f
284-11	V. A. 3. c. 1. c†)	600	100	26,200	26,200	6,580		.000305	f
286-11		700	100	28,350	26,900	2,590		.000885	f
S295-11	V. A. 1. a. 1. c†)	600	400	12,200	12,200	7,650		.000072	f
S296-11		600	300	16,300	16,300	2,960		.000127	f
S294-11		600	200	22,400	22,400	1,880	.00152	.000060	f
S298-11		600	200	23,200	23,200	1,550			f
S243-9		600	100	33,730	32,200	570	.00229	.000115	f

* f = pure fatigue; i = instability.

† Hold-time conditions were changed to 60 seconds at T_2 and 6 seconds at T_1 to more closely coincide with the previous test program. (reference 1).

TABLE IV
(page 2 of 2)

Spec No.	Test Conditions		$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in. /in. -cycle)		Failure Type *
		T (°C)				$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$	
263-10	V. A. 3. a. <u>1</u> . e)	400	41,900	41,900	14,200	.00419	.00013	f
262-10		400	46,200	46,200	6,728	.00457	.00034	f
261-10		400	60,000	60,000	1,095	.00781	.00249	f
267-10		500	30,800	30,800	2,940	.00419	.00081	f
270-10		500	39,600	39,600	820	.0057	.00168	f
268-10		600	17,600	17,600	1,725	.00343	.00152	f
271-10	V. A. 3. a. <u>1</u> . f)	400	48,500	48,500	3,688	.0057	.00140	f
260-10		400	60,440	56,800	314	.0156	.00993	f
269-10		600	16,500	15,600	1,610	.00267	.00123	f
264-10		600	17,250	15,400	870	.00324	.00165	f
S297-11	V. A. 1. a. <u>1</u> . e)	400	31,100	31,100	19,316	.00305	.00020	f
S275-10		400	39,100	39,100	6,150	.00400	.00045	f
S276-10		400	47,500	47,500	2,225	.00724	.0032	f
S273-10		400	54,500	54,500	285	.0152	.0102	i

TABLE V
LIST OF ETCHES

<u>Material</u>	<u>Etch</u>
304 and 304-L	10% oxalic acid in water; electrolytically at 3 volts.
403	3 parts glycerol, 2 parts hydrofluoric acid, 1 part nitric acid.
A302-B	5% Nital
A387-D	10% ammonium nitrate or 2% Nital
Incoloy 800	10% oxalic acid in water; electrolytically at 3 volts.
Zircaloy-2	4 parts glycerine, 4 parts nitric acid, 1 part hydrofluoric acid.

TABLE VI

(page 1 of 2)

TESTS ON A387-D STEEL (CROLOY 2-1/4)

Spec. No.	Test Conditions	$\Delta\sigma_m$	$\Delta\sigma_r$	N	Strain Range, $\Delta\epsilon$ (in./in. -cycle)		Failure Type *	ΔW_s (in. -lb/in. ³)
		(psi)	(psi)		$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$		
479	IV. B. 1. a. <u>1</u> . a)-1c	47,600	74,600	1,186	.00520	.00262	f	185
450	-1d	75,800	65,000	174		.00762	i	
460	-3a	43,530	43,530	2,371	.00319	.00121	f	
452	-4c	73,210	73,210	(103,000 NF)	.00293	.000060	NF	
483	-4d	95,220	95,220	2,122	.00450	.00200	f	
480	-†	78,100	78,100	387	.00797	.00492	f	385
482	IV. B. 1. a. <u>1</u> . c)-1c	53,650	48,770	461	.00843	.00365	i	
467	IV. B. 1. a. <u>2</u> . a)-1bb	62,920	62,920	4,900	.00314	.00094	f	44 387
453	-1cc	72,400	68,960	563	.00857	.00655	f	
466	-2a	40,570	40,570	61,400	.00196	.00047	f	
455	-2aa	54,730	54,730	1,806	.00398	.00173	f	
465	-2b	61,900	58,140	335	.00673	.00480	i	
468	-†	47,410	47,410	4,395	.00310	.00115	f	
456	IV. B. <u>1</u> . a. <u>2</u> . c)-1bb	57,560	52,460	1,934	.00422	.00186	f	
454	-2aa	48,100	42,090	1,618		.00230	f	
474	-2b	54,940	40,760	305		.00624	i	
470	IV. B. 1. a. <u>3</u> . c)-1d	63,750	44,060	328		.00860	i	

* f = pure fatigue; i = instability; NF = no failure.

 $T_2 = 625^\circ\text{C}$; $T_1 = 175^\circ\text{C}$. $T_2 = 625^\circ\text{C}$; $T_1 = 175^\circ\text{C}$.

TABLE VI

(page 2 of 2)

Spec. No.	Test Conditions	$\Delta\sigma_m$	$\Delta\sigma_r$	N	Strain Range, $\Delta\epsilon$ (in. /in. -cycle)		Failure Type *	ΔW_s (in. -lb/in. ³)
		(psi)	(psi)		$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$		
477	IV. B. 1. b. <u>2</u> . a)-**	45,200	45,200	32,421		.000750	f	
478	-2b	58,840	54,300	947	.00667	.00508	f	
463	IV. B. 1. c. <u>1</u> . a)-2aa	54,360	54,360	3,365		.00238	f	
484	IV. B. 1. a. <u>1</u> . e)-††	86,490	86,490	3,065	.00442	.00123	f	
481	-††	94,930	94,930	1,385	.00538	.00235	f	
473	-††	95,800	95,800	1,419	.00781	.00419	f	
451	-††	76,640	76,640	(103,700 NF)	.00308	.000101	NF	
469	-††	75,230	75,230	1,925	.00438	.00178	f	
472	-††	86,400	86,400	802	.00752	.00449	f	
471	-††	82,700	82,700	333	.0104	.00713	f	

** $T_2 = 625^\circ\text{C}$; $T_1 = 375^\circ\text{C}$.

†† $T = 400^\circ\text{C}$.

‡‡ $T = 500^\circ\text{C}$.

TABLE VII

(page 1 of 3)

TESTS ON A302-B STEEL

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in. /in. -cycle)		Failure Type *
					$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$	
295	IV. A. 1. a. <u>1.</u> b)-†	74,130	74,130	493	.00504	.00126	f
262	IV. A. 1. a. <u>2.</u> a)-1c	88,500	88,500	15,215	.00297		f
301	-1c	86,000	86,000	5,450	.00336	.000095	f
278	-2a	46,650	46,650	83,176	.00192		f
253	-2b	76,480	76,480	520	.00536	.00235	f
261	-2c	95,100	60,000	75	.0183	.0173	i
299	-†	75,390	75,390	745	.00341	.00143	f
273	IV. A. 1. a. <u>2.</u> c)-1b	63,120	58,675	28,170	.00231		f
260	-1c	82,640	78,840	1,660	.00306		f
250	-2b	66,640	58,850	200	.00570	.00314	i
251	-2c	70,750	29,000	38	.0189	.0177	i
276	-4b	61,300	57,200	(39,710 NF)	.00211		NF
275	-4c	83,835	80,230	(75,951 NF)	.00288	.000029	NF

* f = pure fatigue; i = instability; NF = no failure.

† $T_2 = 675^\circ\text{C}$; $T_1 = 325^\circ\text{C}$.

TABLE VII
(page 2 of 3)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in. -cycle)		Failure Type *
					$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$	
258	IV. A. 1. a. <u>3</u> . a)-1c	79,400	79,400	19,550	.00376		f
271	-1c	75,000	75,000	21,274	.00295	.000060	f
270	-1cc	90,120	90,120	696	.00462	.00119	f
265	-1d	90,100	86,370	192		.00540	i
256	-2b	72,280	72,280	620	.00427	.00211	f
264	-2c	72,800	48,300	155	.0171		i
279	-†	90,250	90,250	4,369	.00356	.000049	f
285	-‡	89,700	89,700	4,350		.000030	f
284	-†	67,700	67,700	2,677	.00250	.000116	f
254	IV. A. 1. a. <u>3</u> . c)-1c	74,320	71,350	2,740	.00267		f
268	-1c	71,500	68,650	3,960	.00297		f
280	-1c	73,120	70,110	4,010	.00292	.000089	f
283	-1cc	80,940	80,940	167			i
286	-1cc	79,020	73,820	212		.00123	f
297	-1d	83,900	69,700	99		.00885	i
255	-2b	61,600	57,500	355	.00470	.00182	i
281	-†	78,090	74,500	1,333		.000187	f
302	IV. A. 1. b. <u>2</u> . a)-2aa	60,780	60,780	24,575	.00235	.000095	f
263	-2c	75,800	46,700	85			i
296	IV. A. 1. c. <u>1</u> . a)-†	72,760	72,760	1,688		.00172	f
300	IV. A. 1. c. <u>2</u> . a)-1c	79,100	79,100	36,000	.00288	.000033	f
252	-2b	66,000	66,000	1,480			f

† $T_2 = 675^\circ\text{C}$; $T_1 = 325^\circ\text{C}$.

‡ $T_2 = 625^\circ\text{C}$; $T_1 = 175^\circ\text{C}$.

TABLE VII
(page 3 of 3)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in. -cycle)		Failure Type *
					$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$	
267	IV. A. 1. c. <u>3</u> . a)-1cc	81,500	81,500	3,582		.00133	f
269	-1d	82,475	82,475	210		.0050	i
298	-1d	83,900	83,900	188		.00737	i
282	-†	76,500	76,500	(13,060 NF)		.000037	NF
292	IV. A. 1. a. <u>1</u> . e)-**	118,500	118,500	12,700	.0056	.00018	f
293	-**	145,240	145,240	3,519	.00578	.00037	f
291	-**	156,120	156,120	1,067	.00800	.00214	f
290	-**	147,000	147,000	1,512	.00810	.0020	f
304	-**	149,100	149,100	1,042	.00779	.00180	f
303	-**	156,500	156,500	231	.0140	.0073	f
274	-††	82,000	82,000	39,000	.004	.00005	f
294	-††	121,550	121,550	1,833	.0055	.00148	f
287	-††	116,870	116,870	1,850	.00636	.00180	f
289	IV. A. 1. a. <u>1</u> . f)-††	138,700	128,700	102	.0127	.0075	i

† $T_2 = 625^\circ\text{C}$; $T_1 = 175^\circ\text{C}$.

** $T = 400^\circ\text{C}$.

†† $T = 500^\circ\text{C}$.

TABLE VIII
TESTS ON INCOLOY 800

Spec. No.	Test Conditions	T ₂ (°C)	T ₁ (°C)	T _m (°C)	ΔT (°C)	Δσ _m (psi)	Δσ _r (psi)	N	Strain Range, Δε (in./in.-cycle)				Failure Type *	Fracture Path †	ΔW _s (in.-lb/in. ³)
									Δε _{t,l}	Δε _{t,s}	Δε _{p,l}	Δε _{p,s}			
1	IV.C.1.a.1.c)-2cc			500	600	81,800	63,600	50	.0105	.0141	.0069	.0113	i	IC	745
2	-2bb			500	400	81,600	70,500	703	.0067	.0091	.0030	.0058	f,i	TC	366
3	-2cc	850	150	500	700	82,300	62,000	33	.0129	.0172	.0094	.0137	i	IC	922
4	-2bb	750	250	500	500	81,500	66,500	180	.0082	.0101	.0046	.0069	i	IC	445
5	IV.C.1.a.1.† -2bb	750	250	500	500	75,300	53,700	90	.0085	.0114	.0065	.0089	i	IC	70
7	IV.C.1.a.1.a)-2aa	650	350	500	300	75,075	75,075	5,470	.0047	.0047	.00078	.00078	f	TC	541

*. f = pure fatigue; i = instability.

† TC = transcrystalline; IC = intercrystalline.

‡ Hold time = 1 hour.

TABLE IX
(page 1 of 2)

TESTS ON 403 STAINLESS STEEL

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)		Failure Type *	ΔW_s (in.-lb/in. ³)
					$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$		
359	III. A. 4. a. <u>1</u> . a)-1c	70,500	70,500	3,219		.00136	f	81
360	-1c	74,300	74,300	3,675	.00510	.00210	f	120
365	-2b	67,910	64,180	341	.00985	.00750	f	
368	-2b	62,700	62,700	632	.00794	.00460	f	
375	III. A. 4. a. <u>1</u> . b)-1c	78,300	78,300	1,226	.00698	.00392	f	
356	III. A. 4. a. <u>1</u> . c)-1c	66,000	61,000	3,410	.00403	.00180	f	
353	-1d	73,250	61,000	475	.01075	.00805	f	
366	-1d	70,150	54,480	157	.0144	.0116	i	
364	-2b	45,620	37,900	369			f	
370	III. A. 4. a. <u>1</u> . dd)-1c	50,320	50,320	13,282			f	
372	-1c	75,000	75,000	1,650		.00157	f	
371	-2b	62,000	62,000	145		.00140	i	
399	-2b	75,850	64,870	117		.00235	i	
392	III. A. 4. a. <u>2</u> . a)-1bb	70,000	70,000	11,888	.00280	.00020	f	
380	-1cc	81,500	81,500	1,784	.00470	.00160	f	105
379	-1d	74,600	68,200	606	.0105	.00754	f	492
395	-1d	71,200	64,900	658	.0110	.00957	f	477
396	-2a	42,800	42,800	57,203	.00167	.000071	f	
384	-2aa	51,550	51,550	9,500	.00275	.000190	f	

* f = pure fatigue; i = instability.

TABLE IX
(page 2 of 2)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in. -cycle)		Failure Type *	ΔW_s (in. -lb/in. ³)
					$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,s}$		
386	III. A. 4. a. <u>2.</u> a)-2b			853		.00353	f	
387	-2b	46,680	46,680	758	.00660	.00429	f	
393	-3a	41,400	41,400	5,317	.00365	.00152	f	
394	-3a	42,660	42,660	5,259		.00130	f	
391	-4c	72,950	72,950	38,000	.00269	.000063	f	
385	-4d	88,500	88,500	6,056	.00417	.00119	f	
397	III. A. 4. a. <u>2.</u> c)-1bb	59,500	56,700	10,389	.00259	.000190	f	
388	-1cc	69,000	59,000	953	.00885	.00580	f	
401	-1cc	66,760	55,220	649		.00586	f	
390	-2aa	51,500	45,500	2,608		.00104	f	
400	III. A. 4. a. <u>2.</u> c†)-1d	63,500	44,000	425		.0123	f	
383	III. A. 4. a. <u>3.</u> a)-1c	68,500	68,500	14,397	.00255	.00019	f	13.2
355	III. A. 4. a. <u>1.</u> e)-†	88,000	88,000	13,015	.00317	.000245	f	
354	-†	95,400	95,400	8,310	.0044	.00090	f	
358	-†	98,250	98,250	3,887	.00519	.00185	f	
363	-†	105,000	105,000	1,520	.00792	.00380	f	
374	-†	111,375	111,375	437	.0131	.0081	f	
352	-†	124,230	124,230	108	.0158	.0122	i	
377	-†	112,500	112,500	18	.0216	.0186	i	
361	-††	79,000	79,000	2,924	.00423	.00144	f	
369	-††	80,500	80,500	1,026	.00519	.00209	f	
350	-††	79,500	79,500	460	.00846	.00494		

† 1-hour hold time.

† T = 400°C.

†† T = 500°C.

TABLE X
(page 1 of 2)

TESTS ON 304 STAINLESS STEEL
Annealed One Hour at 1038°C and As-Received (Factory-Annealed)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$		
<u>Annealed</u>										
13	I. B. 1. a. <u>1</u> . a)-3a	45,000	45,000	11,300	.0039		.0011	(.0014)	f	TC
26	-3b	55,200	50,000	85	.0071		.0046	(.0104)	f,i	
23A	-3c	57,600	52,000	27	.0095		.0071	(.0177)	i	
10	I. B. 1. a. <u>1</u> . c)-1c	65,100	63,000	4,714	.0067		.0042	(.0073)	f	TC
14	-1d	70,200	62,500	90	.0105		.0071	(.0135)	i	IC
27	-1d	62,000	62,000	40					i	IC
29	-1d	68,400	57,000	35	.0101				i	TC, IC
30	-2c	60,100	47,500	28	.0108		.0085		i	
4	-2d	63,000	43,000	16	.0145				i	TC, IC
6	-3a	45,500	43,000	6,100	.0034				f	TC
43	-3a	46,400	41,900	4,665	.0033		.0014	(.0028)	f	TC
7	-3b	51,500	42,000	89	.0071		.0057	(.0134)	i	IC
12	-3c	50,600	39,100	29	.0089		.007	(.0147)	i	IC
11	-3d	53,700	38,500	21	.0114		.0101	(.0202)	i	
25	I. B. 1. b. <u>1</u> . c)-3b	50,200	40,500	65	.0068		.0047	(.0110)	i	
42	-3c	51,400	38,500	25	.0074		.0052	(.0109)	i	
24	-3d	44,000	36,000	19	.0115		.009	(.018)	i	

* Values in parentheses are calculated from reference 2.

† f = pure fatigue; i = instability.

‡ TC = transcrystalline, IC = intercrystalline.

TABLE X
(page 2 of 2)

Spec. No.	Test Conditions	$\Delta\sigma_m$	$\Delta\sigma_r$	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡
		(psi)	(psi)		$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$		
<u>As-Received</u>										
1	I. B. 3. a. <u>1</u> . a)-2b	71,200	71,200	598	.0069				f, i	TC
23	-2d	74,800	58,000	18	.0154		.0128		i	
19	I. B. 3. a. <u>2</u> . a)-3a	49,000	49,000	9,420	.0027				f	TC
5	I. B. 1. a. <u>1</u> . f)-1a	60,800	60,500	5,825	.0075		.0045		f	TC
8	-2a	69,100	68,500	3,150	.0112				f	TC
9	-2b	84,100	82,000	833	.0112				i	TC

TABLE XI-A
(page 1 of 3)

TESTS ON 304-L STAINLESS STEEL
Annealed - Small Grain

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡	ΔW_I (in.-lb/in. ³)
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}$ *			
49	I.A.1.a.1.a)-1b	50,500	50,500	8,775	.0048		.0022	(.0035)	f	TC	1
54	-1c	60,700	60,700	6,674	.0068		.0039	(.0070)	f	TC	160
104	-1cc	70,400	70,400	2,174	.0086		.0055		f	TC, IC	
59	-1d	71,800	71,800	87	.0104		.0065	(.0124)	i	TC	
38	-3a	44,925	44,925	9,792	.0033		.0015	(.0019)	f	TC	47.4
102	-3aa	53,000	53,000	1,836	.0049		.0025		i	TC	101
39	-3b	50,500	47,700	82	.007		.0037	(.0083)	i	TC, IC	
40	-3c	49,400	41,500	30	.009		.007	(.0175)	i	IC	
61	-3d	53,500	42,550	22	.0104				i		
34	I.A.1.a.1.c)-1c	55,000	55,000	6,320	.0067		.0039	(.0068)	f	TC	
33	-1d	69,000	61,500	75					i		
208	-1d	61,000	53,000	44	.010	.0147	.0068	.0119	i		
56	-3b	44,000	36,000	36	.0072		.005	(.01175)	i		
57	-3c	46,100	33,420	24	.0088		.007	(.0147)	i	IC	278
36	-3d	48,400	33,700	25	.0114		.0102	(.0204)	i		393
63	I.A.1.a.1.e)-1a	99,600	99,600	11,396	.0075				f		
103	-1a	49,800	49,800	10,986	.0075		.0048		f	TC	

* Values in parentheses are calculated from reference 2.

† f = pure fatigue; i = instability.

‡ TC = transcrystalline, IC = intercrystalline.

TABLE XI-A
(page 2 of 3)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡	ΔW_I (in.-lb/in. ³)
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$			
62	I.A.1.a.1.f)-1a	48,000	48,000	13,250	.0075		.0049		f	TC	
52	-1a	53,300	53,300	8,425	.0076		.0049		f	TC	
74	-1c	55,400	55,400	1,015	.0075		.0045		f	TC	
35	-3b	77,000	76,200	56	.015		.0108		f		
38	I.A.1.b.1.a)-1b	48,200	48,200	16,045	.0049		.0022	(.0035)	f	TC	
72	-1d	71,700	71,700	85	.0103		.0068	(.0129)	i	TC	
60	-2b	50,600	50,600	400	.0068		.0042		i	IC	165
71	-2c	60,700	54,400	30	.0106		.0078		i		
76	-2d	56,200	45,300	12	.0144		.0116		i	IC	503
69	-3a	38,600	38,600	10,100	.0032		.0015	(.00187)	f		51
70	-3b	52,300	52,300	90	.0068		.0041	(.0092)	i	TC	183
73	-3c	49,500	41,600	33	.0088		.0067	(.01675)	i		
78	-3c	47,800	42,000	22	.009		.0068	(.0143)	i		
51	I.A.1.b.1.c)-1c	57,000	55,750	5,910	.0074		.0043	(.0075)	f	TC	154
37	-1d	60,800	51,850	52	.0104		.0076	(.0144)	i		
41	-1d	63,550	56,000	63	.0103		.007	(.0133)	i		
50	-3b	49,600	38,561	74	.0071		.0048	(.0113)	i		
77	-3d	48,300	36,000	25					i		
201	I.A.1.a.2.a)-1b	51,350	51,350	17,450	.0042	.0041	.0019	.0012	f		
198	-1c	60,120	60,120	6,220	.006	.0056	.003	.0026	f		
205	-3a	44,850	44,850	10,700	.0029	.0041	.0009	.0013	f		
200	-3b	59,000	59,000	577	.0061	.0075	.0036	.0047	i	TC	
202	-3c	56,080	48,500	120	.0086		.0062		i		

TABLE XI-A

(page 3 of 3)

Spec. No.	Test Conditions	$\Delta\sigma_m$	$\Delta\sigma_r$	N	Strain Range, $\Delta\epsilon$ (in./in. -cycle)				Failure Type †	Fracture Path ‡	ΔW_1
		(psi)	(psi)		$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$			(in. -lb/in. ³)
204	I. A. 1. a. <u>2</u> . c) -1b	48,100	45,000	19,300	.0066	.0039	.0027	.00165	f	TC	
203	-1e	58,000	51,350	7,980	.0066	.0064	.0034	.0036	f		
207	-1d	70,100	60,000	1,150	.0093	.0089	.0055	.0059	f		
199	-3a	40,950	36,600	6,640	.0027	.0024	.001	.0013	f	TC, IC	
197	-3b	52,400	39,900	260	.0064	.0093	.0045	.0072	i		
206	-3c	51,000	39,950	115	.0079		.0060		i		

TABLE XI-B

TESTS ON 304-L STAINLESS STEEL
Annealed - Large Grain

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}$		
181	I. A. 2. a. 1. a)-1b	48,600	48,600	13,400	.0049	.0039	.0028	.0021	f	TC
178	-1c	54,400	54,400	3,810	.0064	.0058	.003	.0042	f,i	TC
166	-1d	59,000	59,000	55	.0101	.0143	.0071	.0111	i	IC
179	-3b	47,800	45,000	82	.0069	.0079	.0052	.0066	i	
184	-3c	43,000	36,650	26	.0085	.0106	.0068	.0071	i	IC
174	I. A. 2. a. 1. c)-1b	48,100	42,000	11,900	.0048	.0036	.0015	.0015	f	
176	-1c	52,600	47,500	3,470	.0065	.0052	.0035	.0034	f,i	TC
180	-1d	59,200	51,400	50	.0100	.0071	.0069	.0059	i	
173	-3a	37,600	34,000	1,850	.0029	.0030	.0017	.0015	i	TC
182	-3b	40,900	32,100	38	.0072	.0124	.0055	.010	i	
183	-3c	40,200	29,000	25	.0085	.0154	.0071	.0123	i	

† f = pure fatigue; i = instability.

‡ TC = transcrystalline; IC = intercrystalline.

TABLE XI-C
(page 1 of 3)

TESTS ON 304-L STAINLESS STEEL
As-Received (Cold-Worked)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡	ΔW_1 (in.-lb/in. ³)	ΔW_s (in.-lb/in. ³)
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$				
119	I. A. 3. b. 1. a)-1b	94,800	94,800	9,087	.0040		.0003	(.00048)	f			
92	-1c	102,000	102,000	1,425	.0063		.0009	(.0016)	f	TC	60.3	
107	I. A. 3. b. 1. c)-1b	85,000	80,200	11,076	.0044		.0002	(.00034)	f			
108	-1c	98,600	87,900	1,516	.0062		.0015	(.0026)	f	TC		
121	-3b	68,700	54,100	71	.0067		.0043	(.0101)	i	IC		
122	-3c	67,000	50,400	39	.0086		.0062	(.0130)	i			
120	I. A. 3. a. 1. a)-1a	65,400	65,400	100,000	.0030	.0038	.00019	.00026	f	TC		
93	-1b	94,000	94,000	8,800	.0042		.0002	(.00032)	f			
134	-1b	90,600	90,600	16,000	.0041	.0043	.0001	.0002	f			
175	-1b	88,000	88,000	9,730	.0043	.0043	.0005	.0004	f			
86	-1c	112,000	112,000	1,950	.0065		.0008	(.0014)	f	TC	57.3	
94	-1c	101,000	101,000	2,530	.0059		.0011	(.0020)	f		74.5	
116	-1c	91,200	91,200	2,471	.0059	.0066	.0015	.0028	f		94.8	195
131	-1c	102,800	102,800	2,470	.0056	.0066	.0013	.002	f		101	140
87	-1d	110,000	110,000	288	.0093		.0036	(.0068)	i	TC		
118	-1d	103,200	103,200	209	.0089	.0116	.0038	.0070	i	TC	296	588
128	-1d	108,500	108,500	151	.0094	.0145	.0042	.0091	i		353	789
100	-2c	102,000	93,400	59	.0098	.0098	.0065		i	IC		
91	-3a	65,000	65,000	6,550	.0037		.0002	(.00025)	f	TC	9.6	
158	-3a	58,000	58,000	10,887	.029	.0039	.00038	.00046	f			
171	-3aa	72,400	72,400	1,630	.0046	.0062		.0034	f	TC		

* Values in parentheses are calculated from reference 2.

† f = pure fatigue; i = instability.

‡ TC = transcrystalline; IC = intercrystalline.

TABLE XI-C
(page 2 of 3)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path ‡	ΔW_1 (in.-lb/in. ³)	ΔW_s (in.-lb/in. ³)
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$				
80	I.A. 3. a. <u>1. a</u>)-3b	68,100	68,100	360	.0067		.0031	(.0070)	i	IC	177	
160	-3b	75,200	71,700	155	.0064	.0130	.0034	.0092	i		193	555
84	-3c	76,900	66,900	60	.0083	.0175	.0051	(.01275)	i			
161	-3c	68,340	58,500	40	.0083	.0175	.0052	.0150	i			
89	-3d	67,500	54,000	42	.0104		.0077	(.0200)	i			
109	I.A. 3. a. <u>1. c</u>)-1b	82,800	77,600	5,374	.0040	.0070	.00044	.0015	f	TC	237	92.5
129	-1b	83,800	82,000	6,820	.0045	.0042	.00024	.00059	f		18	35.9
67	-1c	NA	93,000	1,150	.0074	NA	NA	NA	f			
95	-1c	92,800	87,900	1,560	.0062		.0014	(.00245)	f	IC		
130	-1c	84,400	81,000	1,678	.0061	.0084	.0018	.0041	f		106	257
66	-1d	100,300	86,750	94	.0097		.0057	(.0108)	i		413	
127	-1d	105,600	89,600	122	.0095	.0157	.0049	.0110	i		418	1000
123	-2aa	NA	73,500	2,248	.0046				f	TC		
98	-2b	88,200	77,000	315	.0061		.0021		i	TC, IC		
168	I.A. 3. a. <u>1. c</u>)-3a	60,350	57,000	4,020	.0027	.0045	.00058	.0012	f	TC		
64	-3b	64,800	51,600	105	.0068		.004	(.0094)	i	IC		
97	-3b	67,100	54,200	49	.0065		.0035	(.0082)	i			
99	-3b	61,800	50,600	38	.00651		.0041	(.0096)	i		250	
115	-3b	69,460	53,400	82	.0068	.0143	.0047	.0110	i			
126	-3b	71,320	57,300	115	.0067	.0155	.0043	.0121	i		258	726
65	-3c	66,540	46,900	38	.0086		.0059	(.0124)	i	IC		

NA = not available.

TABLE XI-C
(page 3 of 3)

Spec. No.	Test Conditions	$\Delta\sigma_m$ (psi)	$\Delta\sigma_r$ (psi)	N	Strain Range, $\Delta\epsilon$ (in./in.-cycle)				Failure Type †	Fracture Path †	ΔW_1 (in.-lb/in. ³)	ΔW_s (in.-lb/in. ³)
					$\Delta\epsilon_{t,l}$	$\Delta\epsilon_{t,s}$	$\Delta\epsilon_{p,l}$	$\Delta\epsilon_{p,s}^*$				
106	I. A. 3. a. <u>1</u> . d)-3c	66,700	47,000	36	.0087		.0062	(.0130)	i			
133	-3c	62,400	44,000	38	.0087	.0195	.0067	.0153	i			
162	-3c	62,870	45,750	30	.0087	.0138	.0063	.0103	i		311	521
169	-3c	64,800	47,500	35	.0085		.0062	(.0130)	i		299	
68	-3d	63,300	42,800	32	.0106		.0085	(.0170)	i		413	
117	-3d	65,850	44,600	15	.0102	.0192	.0081	.017	i		402	828
101	I. A. 3. a. <u>1</u> . 3)-1a	NA	115,900	3,728	.0074				f			
81	-1a	107,000	107,000	4,500	.0075		.002		f			
88	-1c	81,500	81,500	990	.0075		.0027		i	TC		
96	-3b	125,900	125,900	44	.0149		.0077		i			
85	-3b	126,400	124,500	21	.015				i	TC, IC		
83	I. A. 3. a. <u>1</u> . f)-1a	105,400	105,400	3,530	.0075		.00175		f			
124	-1b	89,500	89,500	3,500	.0075		.002		f	TC		
82	-1c	79,750	72,000	380	.0075		.0032		i	TC		
125	-2a	109,000	109,000	1,800	.0112	.0112	.0029	.0038	f			
191	I. A. 3. a. <u>2</u> . a)-1b	75,000	75,000	25,900	.0035	.0037	.0001	(.00016)	f			
192	-1c	96,200	96,200	4,980	.0056	.0062	.0007	.0016	f			
194	-3a	54,000	54,000	20,000	.0028	.0020	NM	NM	f			
193	-3b	72,000	72,000	660	.006	.0096	.0023	.0058	f			
195	-3c	72,700	66,000	275	.0077		.004	(.0100)	f			
186	I. A. 3. A. <u>2</u> . c)-1c	86,400	83,500	2,330	.0055	.0083	.0010	.0032	f	TC		
187	-1d	95,800	84,300	335	.0088	.0138	.0035	.0096	i	TC, IC		
190	-3a	54,500	51,000	6,730	.0027	.0023	.0003	.0003	f	TC		
188	-3b	63,100	53,900	220	.0062	.0122	.0038	.0082	i			
189	-3c	66,000	48,600	70	.0079	.0170	.0059	.0144	i			
185	-3d	65,300	46,150	34	.0098	.0202	.0079	.0172	i			

NM = too small to measure.

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