

**TESTING
FOR PREDICTION OF
MATERIAL PERFORMANCE
IN STRUCTURES
AND COMPONENTS**

STP 515



AMERICAN SOCIETY FOR TESTING AND MATERIALS

TESTING FOR PREDICTION OF MATERIAL PERFORMANCE IN STRUCTURES AND COMPONENTS

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presented in two parts;
Part I presented in
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Part II presented at the
Seventy-fourth Annual Meeting
AMERICAN SOCIETY FOR
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Foreword

The Symposium on Testing for Prediction of Material Performance in Structures and Components was presented in two parts; Part I was presented in Anaheim, Calif., 21–23 April 1971 and Part II was presented in conjunction with the Seventy-fourth Annual Meeting of the American Society for Testing and Materials held in Atlantic City, N. J., 29 June–1 July. The symposium was sponsored by the ASTM Committee on Simulated Service and Performance Testing in cooperation with the National Materials Advisory Board and Technical Cooperation Program, Panel P-4. R. S. Shane, National Materials Advisory Board, Washington, D. C., served as symposium and program chairman.

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Introduction

The ASTM Standing Committee on Simulated Service and Performance Testing and the National Materials Advisory Board ad hoc Committee on Testing for Prediction of Material Performance in Components and Structures jointly sponsored a series of four national symposia on predictive testing. There were two objectives of the symposia: (1) to provide a state of the art series of papers on testing aimed at validating design decisions, and (2) to provide a tutorial input to the National Materials Advisory Board ad hoc Committee on Testing for Prediction of Material Performance in Components and Structures. (The report of the Committee was published as NMAB 288 and is available from the National Technical Information Service, Springfield, Va. 22151.)

The papers of the first symposium were published in NBS Special Publication 336 "Space Simulation," U. S. Department of Commerce, National Bureau of Standards, Oct. 1970 (for sale by the Superintendent of Documents, U. S. Government Printing Office, Washington, D. C. 20402, SD Catalog No. C13.10.336 \$5.25). Of the papers presented at Anaheim, Calif., April 1971 and at Atlantic City, N. J., June 1971, fifteen papers, considered to be of lasting importance and most closely related, are included in this volume. Many of the other papers presented at Anaheim and Atlantic City are being published in one of the Society's periodicals—*Journal of Materials and Materials Research and Standards*. It is obvious that the symposia produced an abundance of riches. The fourth symposium was presented at Los Angeles, Calif., June 1972.

It has become apparent as principles of reliability have been applied to design that teaching how to decide when to test and how to test is a gap in engineering education. A basic lesson which emerged from the symposia is that testing which confirms experience without anomaly is a waste. It is only when design is done outside of the set of conditions under which experience without anomaly resulted that testing is imperative. Proceeding without adequate testing under these circumstances is not taking a calculated risk as is often asserted; it is jumping off the end of the dock blindfolded onto unknown rocks or into unknown depths.

The fifteen predictive testing papers in this volume cover a spectrum

which ranges from theoretical considerations of empirical strength through actual demonstrations of late developments in detection of flaws to achievement of reliability by a well designed testing program. These papers are not substitutes for the definitive textbook on the role of testing in the design process. Rather, it is hoped that this volume will not only instruct but will also stimulate the reader to further consideration of this basic element of the engineer's art.

R. S. Shane

Staff Scientist
National Materials Advisory Board, National Research Council
National Academy of Sciences/National Academy of Engineering
Washington, D. C. 20418
symposium general chairman

An Investigation of Fatigue Life Performance in Lap-Type Solder Joints

REFERENCE: Coombs, V. D., "An Investigation of Fatigue Life Performance in Lap-Type Solder Joints," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 3–21.

ABSTRACT: The reliable operation of data processing equipment depends upon the mechanical integrity of many thousands of solder joints, particularly upon the resistance of the joints to strain-cycle fatigue fractures resulting from component temperature (and thermal expansion) excursions. In an attempt to develop life prediction techniques for solder joint fatigue fracture, two test programs were conducted—torsion fatigue tests of pure tin, pure lead, tin-lead eutectic solder, and tin-lead-indium solder, and shear-fatigue tests of lap-type solder joints employing tin-lead eutectic solder. Although solder joints are operated in an "elevated temperature" range (0 to 100 C) for these metals, no significant temperature effect was discovered in the fatigue life behavior of eutectic tin-lead solder. It was observed that increasing temperature enhanced the fatigue life of pure lead (for the cyclic strain rates employed). The fatigue behavior curves provided a design aid for conservative estimates of solder joint lives.

KEY WORDS: predictions, circuit interconnections, soldered joints, failure, fractures (materials), stress cycle, creep properties, stresses, fatigue (materials), crack initiation, crack propagation

Solder Joint Fractures

Reliable operation of data processing equipment depends on the electrical integrity of each of hundreds of thousands of solder joints that interconnect the circuits. However, solder joints of initially high integrity are subject to electrical failures due to mechanical fractures while in service. Although these fractures can result from a variety of loading situations, the most prevalent source is probably cyclic temperature experienced by circuit components during equipment start-up and shut-down periods (heating and cooling).

Figure 1 shows a schematic representation of common solder joint assembly situations which often exhibit solder joint fractures due to temperature cycling. Each situation involves a metal conductor soldered to

¹ IBM Corporation, Components Division, Endicott, N. Y. 13760.

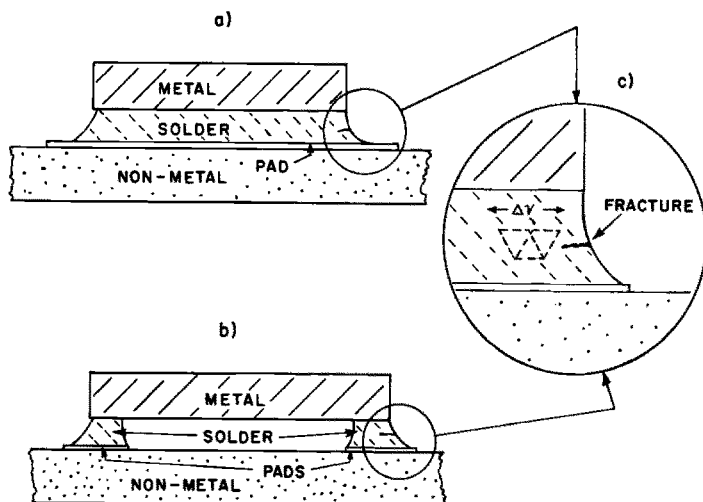


FIG. 1—Representation of solder joint assembly situations.

metal pads (or a pad) which are bonded to a nonmetallic member, such as an epoxy-glass printed circuit board or a ceramic module body. When these solder joint assemblies are subjected to cyclic temperatures, the different rates of thermal expansion and contraction of the metals and nonmetals produce dimensional interferences between the metallic conductor and the nonmetallic member.

If solder were a "rigid" material, such interferences would result in cyclic stresses in the members of the solder joint assemblies. However, conventional tin-lead eutectic solder (63 percent tin, 37 percent lead) creeps readily at and above room temperature (typical creep rate behavior for tin-lead solder is shown in Fig. 2). Therefore, relaxation of

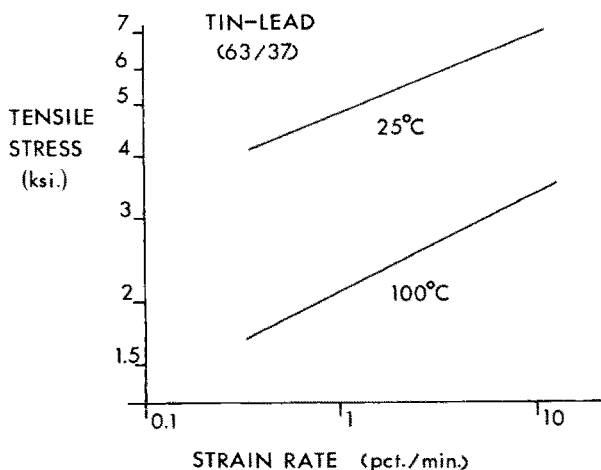


FIG. 2—Typical creep rate behavior for cast tin-lead alloy.

stress (to near zero values) in solder by creep occurs very rapidly in comparison to the rate of heating or cooling of the assembly; no stresses of significant magnitude can be generated or maintained in the assemblies. The consequence is that the cyclic dimensional interferences are accommodated by cyclic plastic (creep) strains in the solder layers, without the presence of significant cyclic stress amplitudes. Solder joint fractures (such as those in Fig. 1) are therefore considered to be strain-dependent failures, or "strain-cycle fatigue fractures."

Life Prediction Approach

The absence of solder joint stress is important as it assures that the creep-rupture type of failure mechanism (which is stress-dependent) will not modify the fatigue fracture behavior and unnecessarily complicate the task of solder joint life prediction. Prediction of solder joint failure lives (numbers of temperature cycles to fatigue fracture) is an ability greatly to be desired; it is valuable in product service performance assurance and in the development of improved solder alloys and solder joint assembly designs.

Our investigation was directed at acquiring the type of fatigue life data on solders which would support solder joint life predictions. A logical approach would be to obtain, by laboratory fatigue testing of solder, the type of data (cyclic strain range versus fracture life, Fig. 3) employed by Coffin[1] and Manson[2]. A fracture life prediction could then be taken directly from such a curve as that shown in Fig. 3, once the anticipated cyclic strain range value had been computed for the action of a cyclic temperature on a specific solder joint assembly.

Several precautions should be carefully considered in this approach.

(1) Accelerated laboratory fatigue life tests should not be operated so rapidly as to not allow stress relaxation by creep to limit solder specimen stresses to near zero levels.

(2) The degree of crack development which is termed fatigue failure

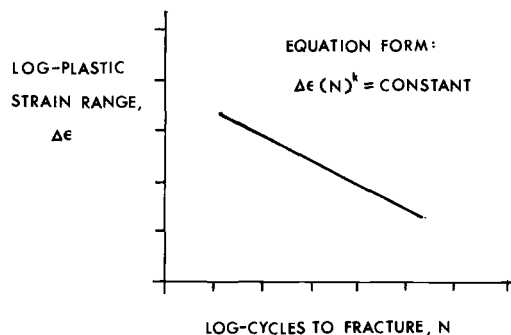


FIG. 3—Conventional manner of presenting strain cycle fatigue behavior.

should be the same in laboratory testing as it would be in a service functional (electrical) joint failure.

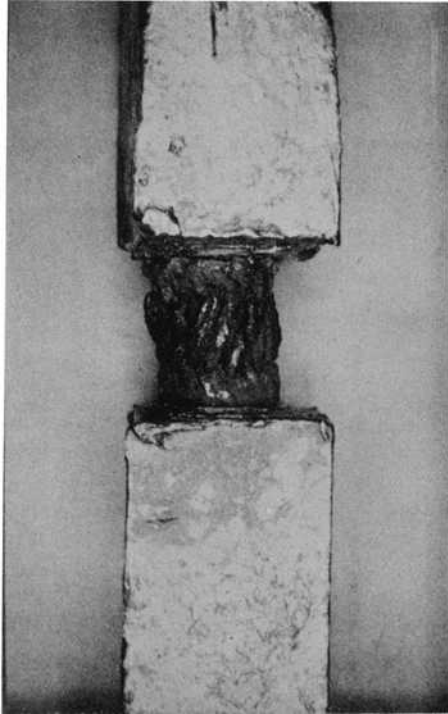
For example, in the solder joint fracture in Fig. 1*a*, it is expected that a crack initiating at the solder layer surface would at first propagate rapidly (due to the strain concentration it produces) and then more slowly, as the length of the "interference couple" (between the metallic conductor and the nonmetallic member) is reduced by the amount of the crack length. In such a case, the crack might never cause complete mechanical fracture (and electrical failure) of the solder joint and might therefore be considered only a "cosmetic failure."

Conversely, the solder joint representation of Fig. 1*b* is a situation where an initiating crack might cause a complete fracture of one of the two solder joints before crack propagation could significantly reduce the length of the interference couple (and thereby reduce the magnitude of the cyclic strain). This second situation could involve a complete functional failure of the solder joint almost immediately after initiation of a crack at the solder surface. Although both situations are encountered in computer circuits, in the interest of conservative life prediction practices, it was assumed that complete functional failure would occur immediately after the initiation of a fatigue crack. It was therefore considered important to insure that the laboratory test methods also produced a situation in which, immediately after initiation, a crack would tend to propagate rapidly to complete fracture (that is, the cyclic plastic strain level increased, not decreased, as the crack propagated).

Fatigue Test Methods

Two types of test methods were used in the investigation—a torsion fatigue test on bulk solder specimens, and a lap shear fatigue test on thin solder layer joint specimens. The torsion fatigue test, which employed a hollow cylinder specimen, permitted examination of the influence of test temperature on solder fatigue behavior. The lap shear, solder joint test permitted examination of the influence of solder layer thickness on solder fatigue at room temperature.

In the selection of a bulk solder test method, tension compression and bending fatigue situations were immediately ruled out, as they did not produce the condition of cyclic shear strain, which occurs in the solder joint types of Fig. 1. Cyclic torsion tests did produce cyclic shear strain and therefore were attractive for the purposes of the investigation. Torsion fatigue tests on solid cylinder specimens did not satisfy the conditions that an initiating crack would be followed, within a few cycles of plastic strain, by complete fracture across the specimen. Figure 4 shows the surface of a solid cylinder specimen from an exploratory test on pure lead. A "ropy" surface appearance has developed from hundreds of ini-



PURE LEAD

TEMPERATURE 25°C

SHEAR STRAIN RANGE 30%

LIFE 560 CYCLES

FIG. 4—Effect of crack initiation, without crack propagation, on surface appearance of torsion fatigue specimen.

tiating cracks which did not propagate toward the specimen axis. In order to insure that initiating cracks would represent a condition of sharply reduced section modulus and would therefore represent an accelerating crack propagation condition, a hollow cylinder specimen was adopted for the torsion fatigue test method.

Figure 5 shows the hollow cylinder specimen and the means in which torsional cyclic strains were applied to it. The following procedure was used for preparing the specimen for test. Specimen blanks (Fig. 5a) were

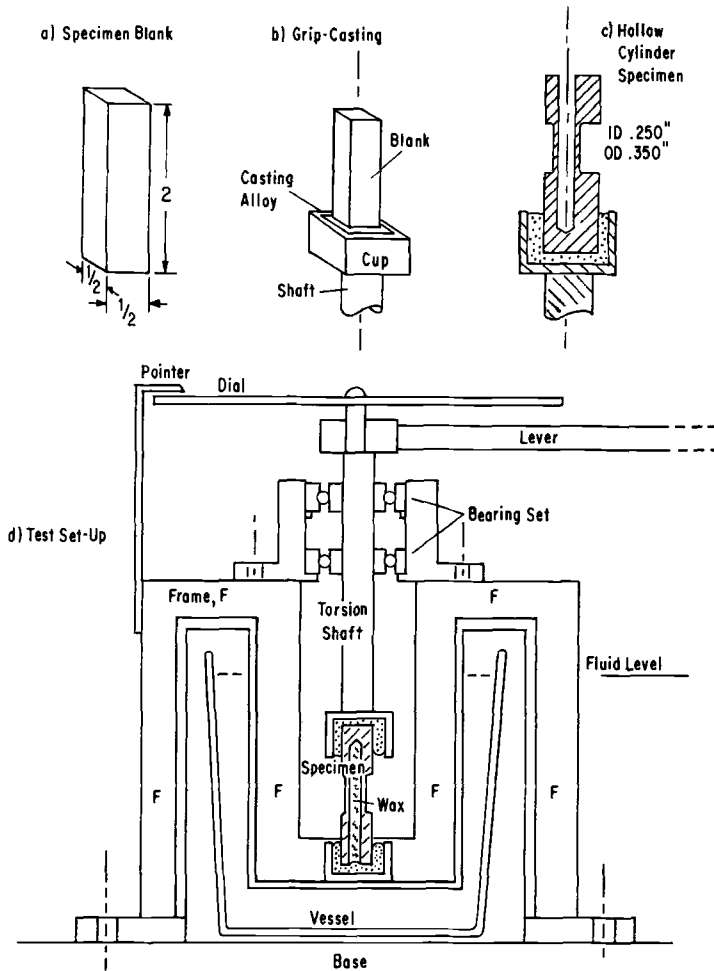


FIG. 5—Sequence of specimen preparation and test setup for torsion fatigue tests.

cast of several solder compositions: (a) tin-lead eutectic (63 percent tin, 37 percent lead), (b) pure tin, (c) pure lead, and (d) tin-lead-indium (54 percent tin, 26 percent lead, 20 percent indium). One end of a specimen blank was "grip-cast" into a square cup at the end of the torsion shaft (Fig. 5b). In each case the grip-casting alloy was a eutectic solder alloy composition of lower melting point than the specimen composition. The purpose of the "cast grip" was to provide specimen end fixation which would not slip or creep during testing. A hole was then drilled axially in the specimen and a reduced outer diameter was machined on the specimen to provide a hollow cylinder test region (Fig. 5c). This grip-casting and machining sequence insured that the test region

"shell" would be concentric with the axis of the torsion shaft which was used to apply a cyclic twist to the specimen.

The specimen and torsion shaft were assembled into the testing fixture (Fig. 5*d*) by slipping the torsion shaft through the bearing set and by grip-casting the other end of the specimen into a fixed cup, employing the same low melting point solder composition as before. The second grip-casting operation insured that the specimen would not be bent during gripping (which would destroy the concentricity of the specimen test region and the torsion shaft). Prior to the second grip-casting operation, the hole in the specimen was filled with candle wax to provide an internal support for the test region which would prevent its collapse toward the axis of the specimen.

Figure 5*d* shows a hand lever and an angular displacement dial that were attached to the top of the torsion shaft. This permitted cyclic twists and torsional strains to be imposed upon the hollow cylinder specimen by a manual operation.² Use of an M-shaped test fixture body (Fig. 5*d*) permitted the test region of the specimen to be immersed in a fluid, contained in the vessel shown, at a temperature desired for the test. Thermocouples embedded in the specimen ends indicated that specimen temperatures remained within 2 to 5 C of the fluid temperature during the tests.

Fatigue test series were conducted at various torsion shaft angular displacements (corresponding to various plastic strain ranges) and at a cyclic frequency of approximately 30 cpm on specimens of the compositions previously listed. Failure was designated as the number of cycles at which a sharp drop in resistance to the cyclic lever displacements could be detected, or "felt" manually. This life criterion was not detectable on solid cylinder specimens at visible crack initiation; on hollow cylinder specimens, however, it was consistently associated with rapid crack propagation and complete specimen fracture, following within a few cycles of visible crack initiation.

Torsion Fatigue Test Results

Fatigue failures obtained in these tests are listed in Table 1 (also included are the curve-fit correlation coefficients); the strain life curves obtained by least squares regression of the data are presented in Figs. 6

² Although manual operation of the torsion fatigue test fixture undoubtedly introduced some "scatter" in the test results, care was taken to avoid a human factor bias in the results. Operator recognition of specimen identifications, and operator anticipation of failure patterns were avoided. The operator instruction was to produce cyclic angular displacements to the "spring-back" condition—that is, it was intended that the specimen strains calculated from the spring-back angular displacements of the dial would reflect plastic strain range values, rather than total strain range values. The wave shape produced by the manual operation was assumed to be approximately a "saw tooth"; the strain rates resulting were estimated to be in the range of 1.5 to 28 percent/s. It was assumed that over this range variations in strain rate would not significantly influence the low cycle fatigue results.

TABLE 1—*Torsion fatigue test data on eutectic tin-lead (63 percent tin, 37 percent lead), tin, lead, and tin-lead-indium alloy (54 percent tin, 26 percent indium, 20 percent lead).*

Alloy 1	Specimen No.	Test Temp., deg C	Cyclic Shear Strain Range, $\Delta\gamma$, percent	Cycles to Complete Fracture	Fatigue Curve Exponent K	Correlation Coefficients
Tin-Lead	0-22	0	7.1	58	-0.65	0.924
	0-25	0	3.5	112		
	0-26	0	11.0	18		
	0-27	0	3.9	68		
	0-17	25	13.9	6	-0.68	0.975
	0-18	25	7.2	24		
	0-19	25	3.4	58		
	0-20	25	2.3	61		
	0-66	25	10.1	10		
	0-63	25	2.7	77		
	0-49	60	7.2	23	-0.55	0.766
	0-44	60	4.9	78		
	0-45	60	3.2	85		
	0-50	60	8.3	45		
	0-71	80	7.9	21	-0.67	
	0-70	80	11.5	12		
Pure Lead	0-13	-195	10.8	86	-0.54	0.987
	0-16	-195	6.9	153		
	0-23	-195	5.2	263		
	0-61	-195	3.9	535		
	0-51	-110	11.3	100	-0.46	0.858
	0-52	-110	5.2	420		
	0-53	-110	9.3	225		
	0-54	-110	7.5	150		
	0-9	25	13.4	102	-0.26	0.911
	0-10	25	9.0	328		
	0-11	25	12.2	340		
	0-12	25	16.4	53		
	0-33	25	11.0	222		
	0-24	25	8.19	603		
	0-47	60	13.3	185	-0.11	0.750
	0-48	60	11.8	257		
	0-14	60	11.2	970		
	0-15	60	14.3	230		
Tin	0-36	25	11.1	291	-0.52	0.941
	0-37	25	9.0	343		
	0-38	25	15.2	221		
	0-39	25	28.0	37		
	0-40	25	21.2	126		
	0-41	25	8.8	351		

TABLE 1 Continued

Alloy 1	Specimen No.	Test Temp., deg C	Cyclic Shear Strain Range, $\Delta\gamma$, percent	Cycles to Complete Fracture	Fatigue Curve Exponent K	Correlation Coefficients
Tin (Cont.)	0-69	100	7.1	295	-0.61	0.547
	0-67	100	11.2	246		
	0-68	100	8.5	373		
Tin-Lead Indium	0-55	25	11.3	9	-0.43	0.999
	0-56	25	7.5	21		
	0-57	25	4.7	63		
	0-58	25	3.4	142		
	0-72	100	7.2	13	-0.59	0.993
	0-73	100	5.5	27		
	0-74	100	2.4	107		
	0-75	100	1.5	183		

through 9. Figure 10 shows photographs of both an untested torsion fatigue specimen and failed specimens with typical fracture appearances.

A significant observation from these results is that the fatigue behaviors of tin-lead solder and of pure tin were not dependent on test temperature; the tentative conclusion was made that failure was fatigue fracture not influenced by creep rupture processes. Apparently, the cyclic frequency (and associated strain rates and shear stresses) was not so high as to induce creep-rupture "damage." A positive dependence of the fatigue life of pure lead and a negative dependence of tin-lead-indium on test temperature are apparently revealed by the curves of Figs. 8 and 9. Interpretation of these temperature dependencies has not been pursued as yet.

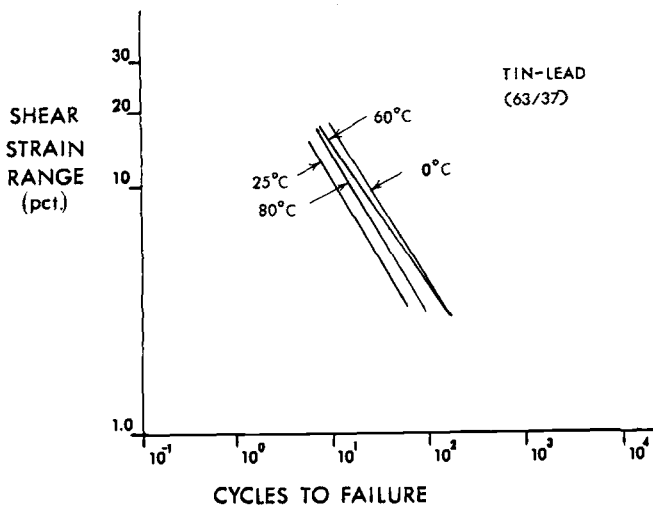


FIG. 6—Strain cycle fatigue behavior of tin-lead solder (63 percent tin, 37 percent lead).

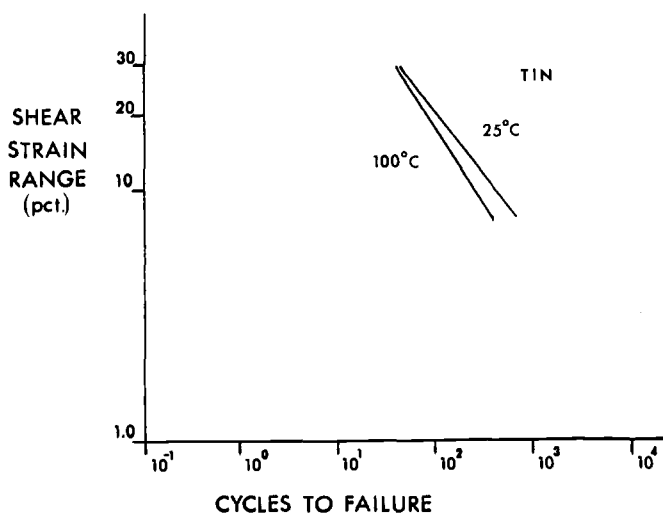


FIG. 7—Strain cycle fatigue behavior of pure tin.

Solder Joint Test Method

The second phase of the investigation—testing of lap-shear solder joints—arose from the concern that the thickness of the solder layer might influence the fatigue life of the solder for any given cyclic plastic strain range. Additionally, the results from fatigue tests of solder joints might provide correlation with the fatigue behavior illustrated in Fig. 6, increasing confidence in the use of both sets of fatigue data in the prediction of solder joint lives. Figure 11 shows the type of specimen selected for the solder joint fatigue test series.

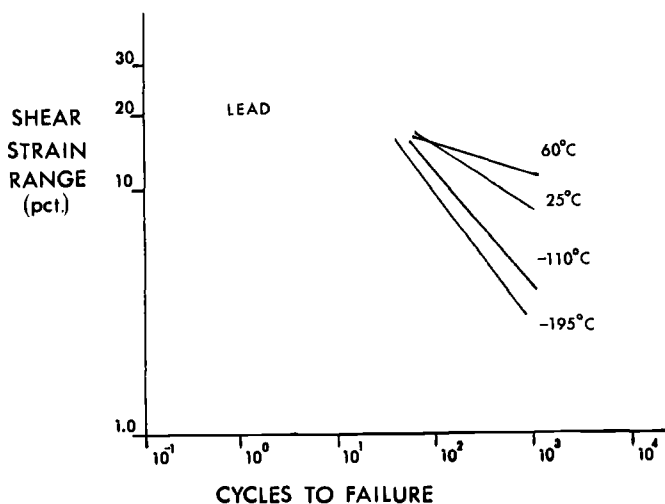


FIG. 8—Strain cycle fatigue behavior of pure lead.

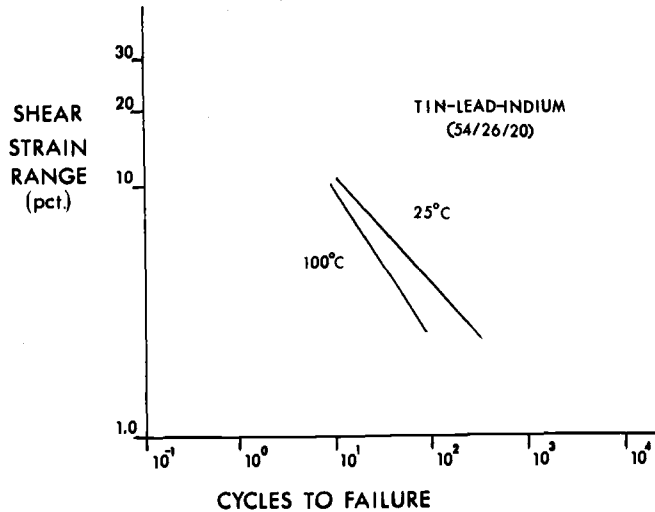


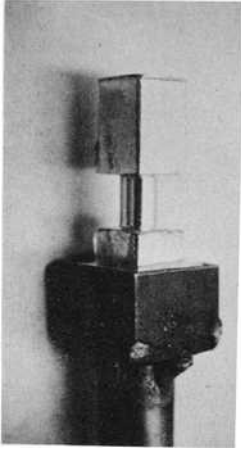
FIG. 9—Strain cycle fatigue behavior of tin-lead-indium solder (54 percent tin, 26 percent lead, 20 percent indium).

Specimens were prepared by prepositioning the corners of two copper strips (0.020 in. thick, by $\frac{5}{32}$ in. wide, by $\frac{5}{8}$ in. long) so that an overlap of about 0.050 in. was produced, with a separating gap in the range 0.001 in. to 0.015 in. This gap was filled with solder by applying a hot soldering gun tip and rosin-core solder wire (0.015 in.-diameter) to the gap region for a few seconds.³ The flat faces of each solder joint specimen were sanded after soldering to provide a smooth profile of the joint region. On some specimens, the fillets were removed (indicated with dashed line in Fig. 11) by the use of a jewelers file.

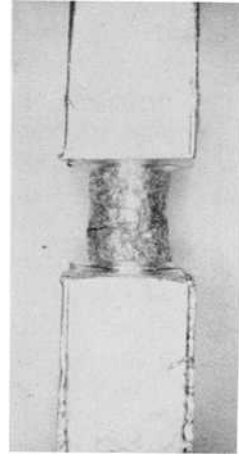
It was initially intended that about a dozen of these specimens with 0.001 in. to 0.015 in.-solder layer thicknesses would be tested under various levels of cyclic plastic shear strain. The test series, therefore, would either produce fatigue behavior correlating with that of Fig. 6, or it would reveal any effect of solder layer thickness which prevented a correlation. However, the specimens with solder layers less than 0.005 in. thick suffered a high mortality rate in preparation, and the test series involved a predominance of specimens with solder layer thicknesses in the neighborhood of 0.005 in.

Two factors influencing the design of the solder joint testing device were the necessities to (1) produce rather small cyclic displacements and (2) observe the crack initiation and propagation along the solder layer

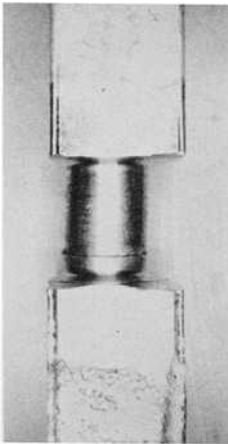
³ Some limitation of temperature (to near the solder-melting point) was effected by feeding solder wire into the joint as fast as it would melt until the gun tip was withdrawn. Metallographic examinations indicated that the time-temperature experience produced insignificant thicknesses (about 80 μ in.) of Cu_6Sn_5 intermetallic compound at the solder-copper interfaces.



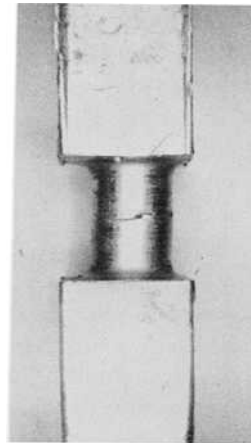
SPECIMEN AND
TORSION SHAFT



TIN-LEAD
 $T = 60^{\circ}\text{C}$
 $\Delta\gamma = 7.2\%$
 $N = 23$ CYCLES



TIN
 $T = 25^{\circ}\text{C}$
 $\Delta\gamma = 11.1\%$
 $N = 291$ CYCLES



TIN-LEAD-INDIUM
 $T = 100^{\circ}\text{C}$
 $\Delta\gamma = 5.5\%$
 $N = 27$ CYCLES

FIG. 10—Appearances of failed torsion fatigue specimens (with an untested specimen).

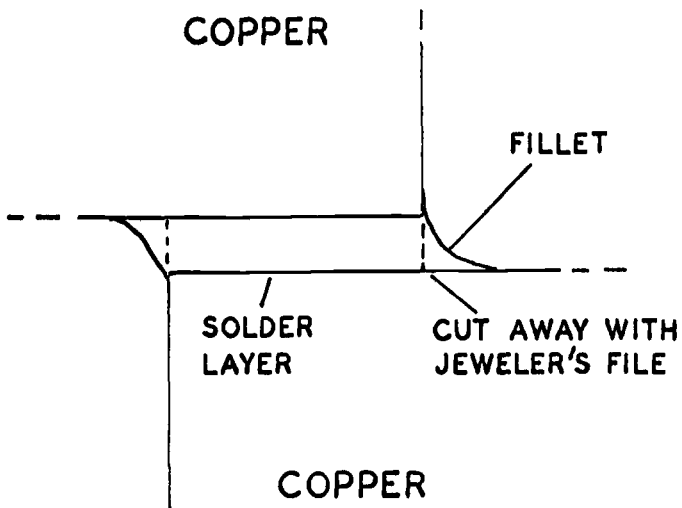
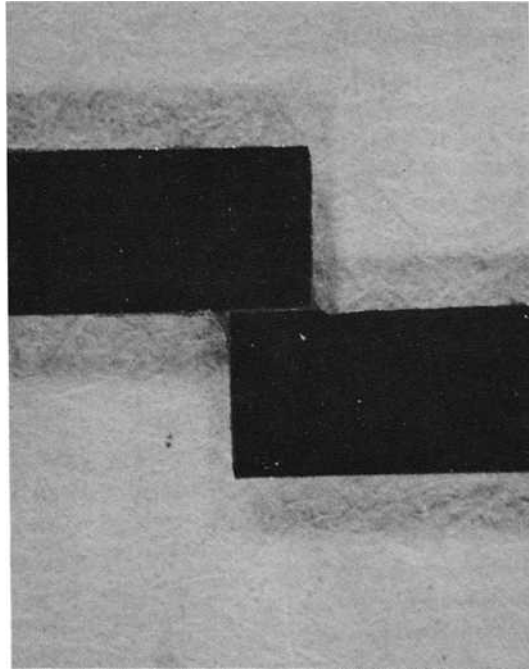


FIG. 11—Lap shear solder joint fatigue specimen ($\times 6$).

profile. It was anticipated that the life range convenient to test would be approximately 10 to 100 cycles and that this would correspond to a plastic shear strain range of 10 to 20 percent. For a solder layer thickness of 0.001 in., and a desired shear strain range of 10 percent, it would therefore be necessary to produce with reasonable accuracy cyclic shear displacements of the magnitude of 0.0001 in. In addition, the question of whether or not an initiating crack would indeed propagate to complete fracture (within a few cycles of strain, in this type of specimen) made it desirable to observe the first appearance of the solder layer profile crack and its propagation to complete fracture. In view of these two desired features, and in the interest of employing a testing scheme of low technical sophistication, the rolling wire, thermal dilation fatigue tester was devised (Figs. 12 and 13).

The device employs the small dimensional changes produced in two stainless steel rods, by heating and cooling of the rods, to produce cyclic displacements of a moving clamping base with respect to the stationary clamping base. This displacement is used to effect a cyclic, lap-shear deflection in the solder joint specimen, which is clamped at the ends to the moving clamping base and to the stationary clamping base. Cyclic thermal dilation of the rods is produced by electrical current pulses passed through the rods in series. Periodic electric heating and expansion of the rods is interrupted by periods of air cooling and rod contractions.

The heating and cooling periods (and in turn the displacements and resulting solder layer shear strain) are controlled by a rolling wire switch, which is activated by the motion of the moving clamping base relative to the stationary clamping base. The moving clamping base rests on a 0.004

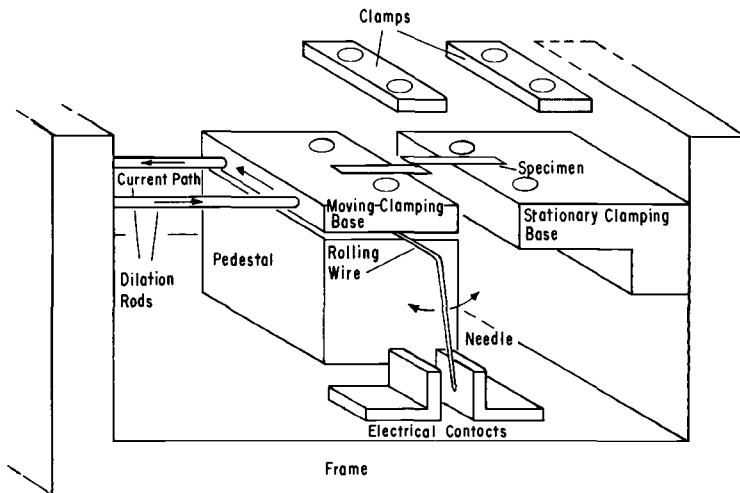


FIG. 12—Schematic diagram of the thermal dilation fatigue tester.

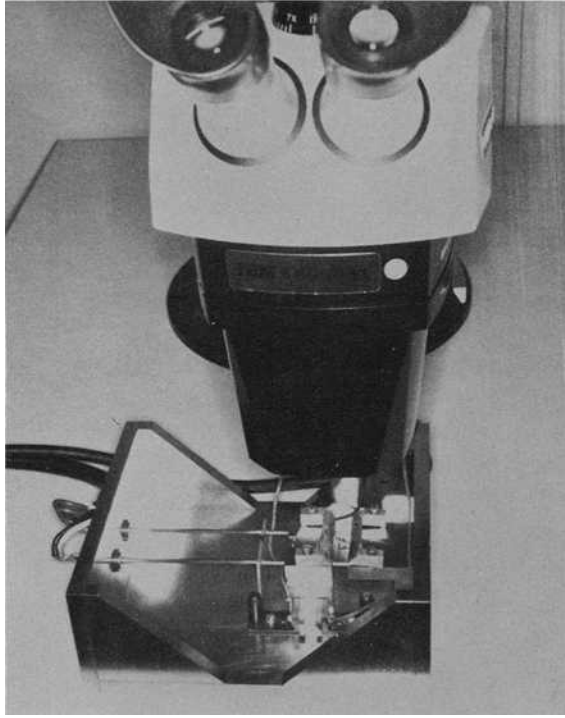


FIG. 13—*Thermal dilation fatigue tester.*

in.-diameter tungsten wire, which in turn rests on a pedestal (Fig. 12). Any motion of the moving clamping base with respect to the pedestal (and to the stationary clamping base) results in a rolling, or angular displacement, of the wire. A section of the wire is bent at a right angle to the rolling section, forming a needle which displays the same angular displacement as the rolling wire. The appreciable motions of the end of the "needle" are a direct mechanical amplification of the very small motions of the moving clamping base and were used as a means to preset and control the moving clamping base motion.

Amplification is expressed by a factor, S/x derived from the expression:

$$x = 2D \sin^{-1} (S/2R)$$

where

- x = moving clamping base displacement,
- D = rolling wire diameter,
- S = displacement of the needle end, and
- R = length of the needle.

For the tests of this investigation the amplification was approximately 129. The preset spacing of two electric contacts (Fig. 12), which could be

energized by the needle end, determined the magnitude of the cyclic displacement of the moving clamping base. The needle end was employed as an electrical trigger (in conjunction with an electronic relay) to turn on and turn off the electrical heating power to the stainless steel rods. The electrical contact spacing therefore determined the cyclic shear strain range (total) in the solder joint specimen. The power supplied for dilation rod heating was approximately 100 W, and the lowest rod temperature value in a cycle was over 100 C. The heating and cooling curves for the rods were assumed to be approximately linear, and the resulting cyclic displacement wave shape was assumed to be a "saw tooth" for these conditions.

As shown in Fig. 13, the device was small and lightweight and could easily be employed under a binocular microscope, so that the progress of a solder layer crack could be visually followed. Calculations and measurements were made to estimate the accuracy of this tester in this investigation. Although a discussion of these is not within the scope of this paper, they indicated that the total strain range values recorded for the tests were accurate to about five percent. In particular, the calculation of a time constant (for relaxation of stress to 37 percent of its initial value) indicated that the total strain range values could be assumed to represent the plastic strain range values without introducing greater than five percent error. The nominal cyclic displacement rate (for a saw tooth wave shape) was approximately 0.00005 in./s. The resulting strain rates were estimated in the range of 0.4 to 1.35 percent/s.

Solder Joint Fatigue Test Results

Fatigue data obtained on tin-lead solder joints with this test method are listed in Table 2 and plotted in the strain-life graph of Fig. 14. Typical fracture behavior on the solder-joint specimen is illustrated in Fig. 15. The crack usually initiated at one or both points where the solder fillet leg was short. In Fig. 15, two such cracks have propagated toward one

TABLE 2—*Lap-shear fatigue test (conducted at 25 C) data on eutectic tin-lead (63 percent tin, 37 percent lead) solder joints.*

Specimen No.	Solder Layer Thickness, in.	Cyclic Displacement Range, in.	Cyclic Shear Strain Range, percent	Cycles to Crack Initiation	Failure to Life, Cycles to Complete Fracture	Fatigue Curve Exponent K	Correlation Coefficient
SL-1	0.004	0.00046	11.5	16	102	-0.36	0.86
SL-3	0.0056	0.00046	8.2	75	212		
SL-4	0.0089	0.00139	15.6	24	60		
SL-5	0.0127	0.00139	10.9	20	103		
SL-8	0.0063	0.00139	22.1	4	35		
SL-10	0.0050	0.00046	9.2	20	258		
SL-11	0.0051	0.00033	6.5	302	548		
SL-12	0.0037	0.00046	12.4	62	350		

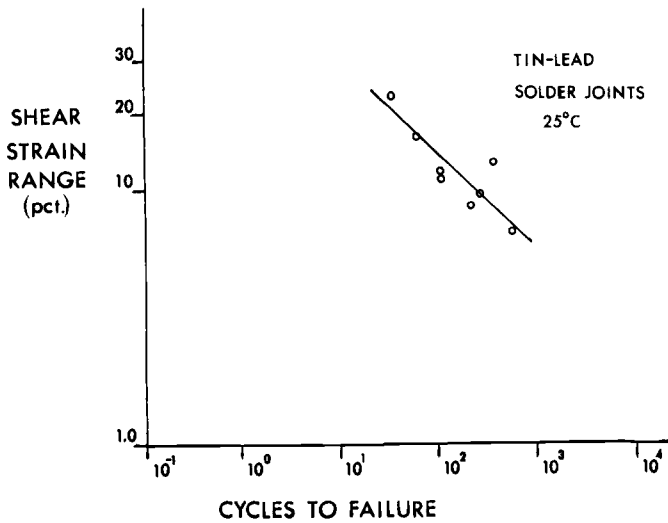


FIG. 14—Results of lap shear fatigue tests on tin-lead solder joints.

another, until a tearing-through of the separating solder layer occurred. The cutting away of the fillets (as shown in Fig. 11b) on some specimens did not have an appreciable effect on joint life, nor did the presence of cracks observed in some specimens before testing was begun. (These were suspected to be due to handling during specimen preparation.) These “early” cracks were not observed to propagate rapidly until at least 75 percent of the joint life had elapsed. The value of crack initiation was therefore questionable as a failure criterion; and the number of cycles

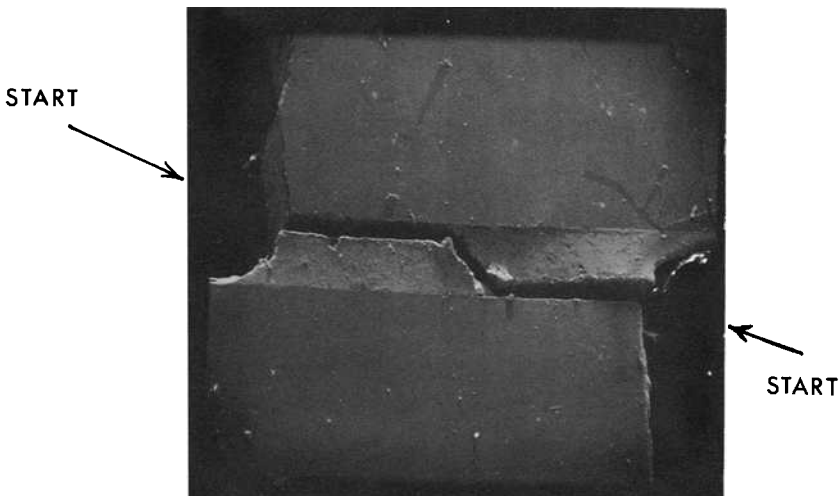


FIG. 15—Typical fatigue fracture appearance on tin-lead solder joints ($\times 44$).

to complete fracture was taken to be "failure" for these solder joint specimens.

Discussion of Results

Comparison of the curve of Fig. 14 with that of Fig. 6 suggests that considerable additional testing will have to be done to establish a satisfactory correlation of results between bulk material specimen tests and solder joint tests. For the immediate purposes of estimating solder joint fatigue lives in service, the curves of Fig. 6 and Fig. 14 are replotted in Fig. 16 as "best case" and "worst case" limit curves. Obviously, Fig. 16

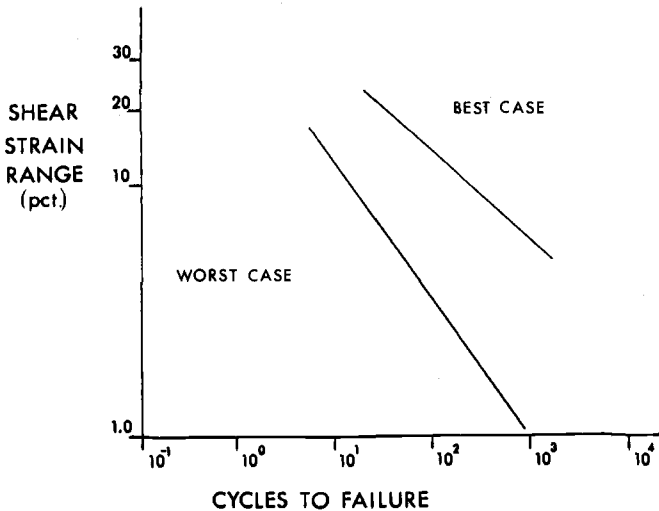


FIG. 16—Envelope of fatigue behavior for tin-lead solder.

is intended as a very tentative design aid. Hopefully, a more systematic investigation of the separate influences of solder layer composition and thickness will improve the correlation between test methods, and thereby lend confidence that the laboratory test data do reflect service behaviors accurately.

Although interpretations of the results in terms of metallurgy or fracture mechanics is beyond the scope of this paper, several observations deserve emphasis. Solder joint assembly situations and service experiences usually involve low strain application rates and rapid stress relaxation. Therefore, the effects of stress, and of interactions of fatigue and creep rupture under conditions of cyclic strain, may not be as important a consideration as in the use of other metals in other types of applications at high temperatures.⁴ A freedom from the effects of creep rupture appears to be sup-

⁴ Application of solder at room temperature and above must be considered "high temperature service," as such temperatures represent at least half the melting point of a soft solder on the absolute temperature scale.

ported by the data of this investigation. The slopes (K) of the curves of Fig. 6 (and listed in Table 1) are in the range (0.5 to 0.7) typical of strain cycle fatigue behavior, and are independent of temperature. The coefficients of correlation listed in Table 1 indicate, in general, good fits of the fracture life data on tin-lead solder to the power relationship recommended by Coffin[1].

The author emphasizes that such observations may be valid only where a solder joint assembly is sufficiently "stiff" to allow rapid stress relaxation and where the initiation of a crack tends to increase the effective cyclic strain range imposed, thereby causing "precipitative failures."

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Predictive Testing in Elevated Temperature Fatigue and Creep: Status and Problems

REFERENCE: Coffin, L. F. and Goldhoff, R. M., "Predictive Testing in Elevated Temperature Fatigue and Creep: Status and Problems," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 22-74.

ABSTRACT: The paper selects two well recognized problem areas of high temperature material behavior, namely, creep and fatigue, and considers those factors which contribute to the prediction of the useful service life of the component. The information available is obtained almost entirely in a laboratory environment, in which simplified load conditions and specimen geometries and controlled environments are employed for material evaluation. We do not consider transferring this information to actual design, other than to identify the importance of such problems as the analytical determination of stresses and strains in the structure, the effect of highly complex loading, or the behavior in a complex environment, but we will draw on several practical problems occurring in high temperature power generating machinery as examples of the topics we wish to discuss namely, creep, fatigue, and creep-fatigue interaction.

The topics of creep and fatigue are considered separately, the discussion being built around three central topics: (1) real problems in prediction, (2) complicating instabilities including the effects of environment, deformation, and microstructure; and (3) some current predictive methods. Consideration of the less well defined problem of creep-fatigue interactions is also included.

KEY WORDS: predictions, stresses, strains, creep properties, creep rupture strength, fractures (materials), crack initiation, crack propagation, failure, grain boundaries, notch sensitivity, shear properties

Nomenclature

Creep and Rupture

ϵ	Strain
t	Time
$\left. \begin{matrix} \sigma \\ S \end{matrix} \right\}$	Stress (uniaxial unless otherwise specified)

¹ Metallurgy and Ceramics Laboratory and Materials and Processes Laboratory, respectively, General Electric Co., Schenectady, N. Y. 12301.

σ	Effective stress =
	$\frac{1}{\sqrt{2}} [(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2]^{1/2}$
$\bar{\epsilon}$	Effective strain =
	$\frac{\sqrt{2}}{3} [(\epsilon_1 - \epsilon_2)^2 + (\epsilon_2 - \epsilon_3)^2 + (\epsilon_3 - \epsilon_1)^2]^{1/2}$
σ_{MAX}	Maximum principal stress
Q	Activation energy
K_T	Elastic stress concentration factor
K_σ	Plastic stress concentration factor
K_ϵ	Plastic strain concentration factor
E	Young's modulus of elasticity
E_s	Secant modulus
ϕ	Function notation such as $\phi(\epsilon)$ or $p(\sigma)$
θ	
$()$	
Σ	Summation
a, b, C, T_A, t_A, K, M	Constants

Fatigue at Elevated Temperature

N, N_f	Number of cycles, cycles to failure
N_i, N_p^e, N_p^e	Number of cycles for initiation, propagation in plastic and elastic regimes
$\Delta\epsilon_e, \Delta\epsilon_p, \Delta\epsilon$	Elastic, plastic and total strain range
$\Delta\sigma, \Delta S$	Actual and nominal stress range, psi
ϵ_f	Tensile ductility
C	Crack length in.
ν	Frequency, cpm
t_h	Hold time, min
T	Temperature, deg F
E	Elastic modulus, psi
$A', \beta', k_1', A, n, k_1, C_2, k_2, \beta$	Coefficients of high temperature fatigue equations
P	Larson-Miller parameter
K_f, K_t	Fatigue and theoretical stress concentration factors
r	Notch root radius, in.
c, a	Constants

The fundamental laws, leading to broad generalizations on which sound engineering analysis and design depend, must be preceded by a thorough understanding of the behavior of materials. The broader this

base, the greater the possibility for the engineer to choose reliable design criteria and select economical and reliable component materials.

Elasto-mechanics has developed in this way, but not the area of plastic behavior. The addition of time and temperature complicates the plastic and elasto-plastic mechanics, and many of the present limitations in analysis and design at elevated temperatures stem from a lack of understanding of materials behavior under these conditions. It would be ideal to understand material behavior; first, in simple high temperature tests; second, how these relate to more complex laboratory tests; and, finally, how these latter tests might be related to actual field service.

Components operating in machines at elevated temperature seldom involve purely fatigue or purely creep behavior; and, generally, the most important problems involve the complex and poorly understood interaction between the two. One can, however, suggest practical situations in which each of these phenomena predominates. For example, in the large turbines used in central station power generation, the massive valves controlling the inlet of high pressure, high temperature steam are limited by creep phenomena and are less affected by stopping and starting cycles. On the other hand, the hot gas path blading in the stationary gas turbine and the aircraft jet engine present problems in elevated temperature fatigue and are limited in design by fatigue and fatigue-creep interactions. Different kinds of materials behavior testing, analytical techniques, and predictability can be expected for given situations.

The designer concerned with such problems may ask significant questions:

1. Can stresses and strains be predicted analytically?
2. Can these stresses and strains be reduced?
3. Are the stresses and strains damaging so as to produce cracking?
4. What life can be expected?
5. What materials should be used?

Each question is another stage in the prediction process. Success in the process depends on the ability to evaluate material response to measurable and controllable testing variables and to determine how they are limiting. It follows that prediction can only be as good as this approach will allow.

Practical predictive limitations are imposed by material response to complex loading, complex geometry, and uncertain histories, among others. A major problem faces us initially in our inability to reproduce given materials in a constant and predictable structural state. Natural inhomogeneities and defects are inherent in all commercially produced materials, and manmade defects (for example, welds), are a necessary condition of our technology. It is beyond the scope of this report to deal with this subject, but the effect on materials properties, design allowances, and the need for statistical treatments is to pose more difficulties for predictive testing.

This report will review relevant material behavior knowledge, limitations imposed by real problems, and predictive techniques. Guidelines for future research and development effort in predictive testing are also presented.

Creep and Rupture

Real Problems in Life Prediction

In many respects design requirements are almost totally those of prediction. Generally, the designer attempts to predict the stresses, strains, and temperatures associated with a component design and then tries to express them in analytical forms. Materials data, usually obtained from simple laboratory tests are then applied in an effort to assess the serviceability of the proposed design and to estimate the life. Both component analysis and the obtaining of materials data in creep-rupture have presented difficult problems.

Figure 1 presents a simple schematic overview of the creep-rupture problem while introducing complicating factors. In the area of analysis every design will be based on limits imposed by creep or rupture, and a prediction as to whether stress or strain criteria apply must be made. Traditional design has been concerned with creep and attention has been

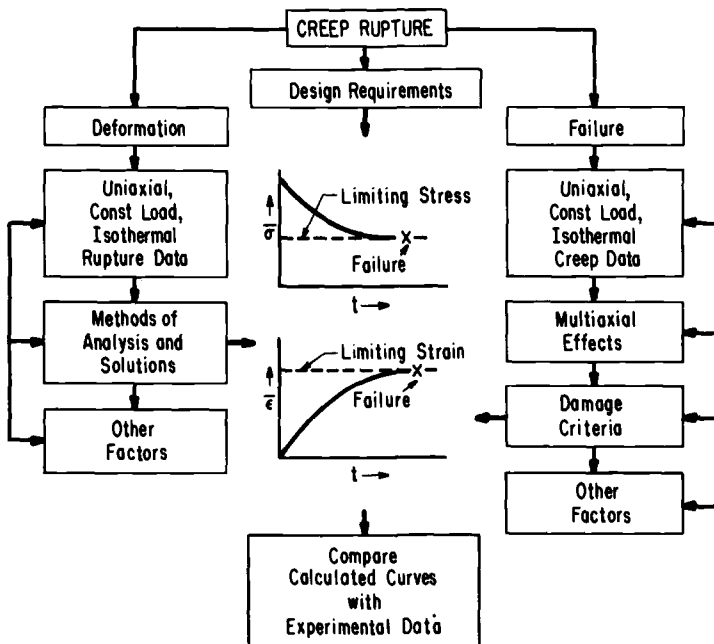


FIG. 1—Creep rupture problem elements.

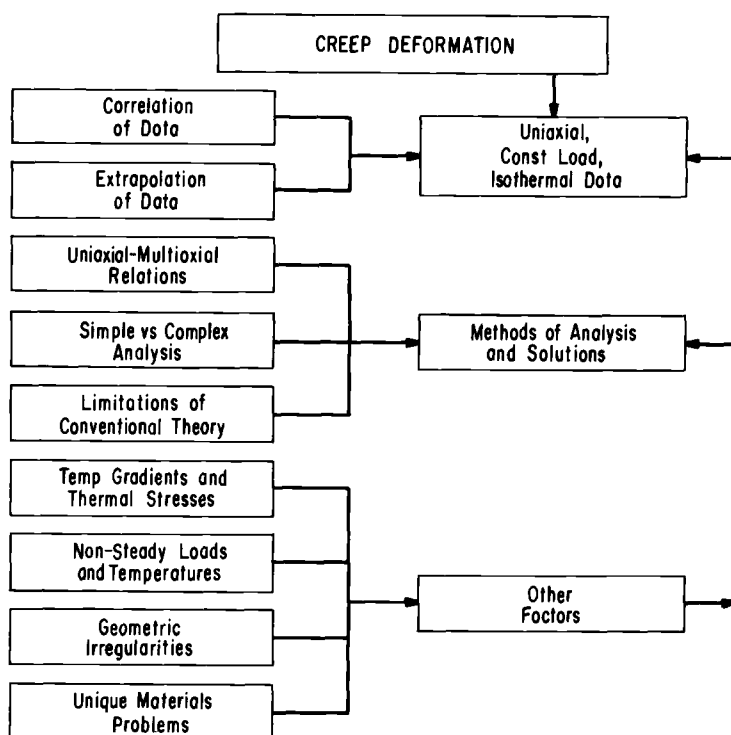


FIG. 2—Details of creep deformation problem.

given to the methods for predicting creep strain, particularly under multiaxial stress conditions. With the advent of highly creep resistant materials, rupture has tended to become the limiting factor in design, and as we shall see, much less information is available for prediction.

In Fig. 2 the creep deformation portion of the previous figure is expanded. The prediction of creep deformation requires a knowledge of materials response which is applied to flow rules for problem solution, that is, the development of a relationship between stress, strain, time, and temperature. Despite a large amount of work, this functional relationship is not generalized. To determine deformations in components subject to multiaxial stresses, an effective stress-strain relationship is assumed based on correlations obtained in uniaxial laboratory tests. The problem is to find a correlation between uniaxial and multiaxial behavior in the plastic creep range. Many solutions have been proposed and good results have been obtained in special cases, yet to date no generalized flow law exists. Further, complications exist as shown in Fig. 2 (other factors).

When fracture is considered the problems become even more complex as shown schematically in Fig. 3. Fracture involves initiation and propagation of cracks and no general method for predicting rupture is accepted. Some of the current methods in use for the prediction of rupture life are

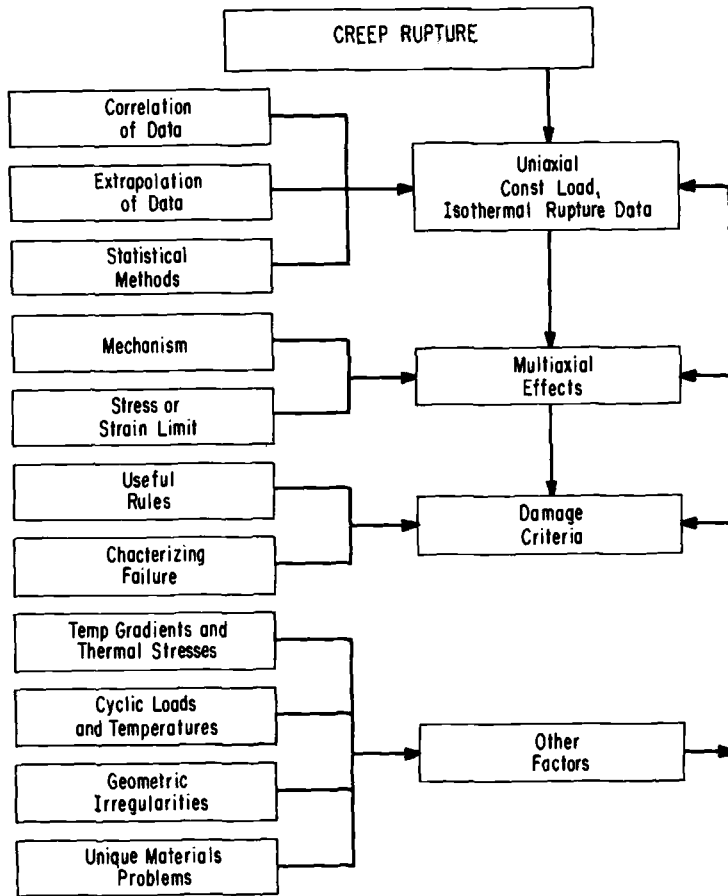


FIG. 3—Details of creep-rupture problem.

discussed later with emphasis on the need for a general rupture criterion.

During the service life of a component operating under essentially steady-state conditions at elevated temperature some changes in stress or temperature or both will occur, even if these are due only to stop-start operations. Damage is incurred under such conditions and the prediction of life for components will depend on the assumption of some rule as to how damage accumulates. Only simple rules are in use today and experimental work, particularly involving multiaxial behavior, is needed to clarify our concepts of damage.

Most elevated temperature failures occur because of geometric irregularities or notches. The condition is one of local stress and strain concentration, which under creep conditions change continuously with time. Here the failure limiting conditions may be the accumulation of effective strain as the concentrated stress is relaxing (lower part of Fig. 1) and hence be related to the material ductility. On the other hand the problem

may be in the inability of the material to relax sufficiently and hence be stress related (upper part of Fig. 1). This problem is perhaps the most practically difficult problem in the design of components at elevated temperature.

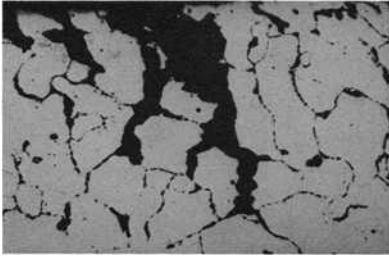
Complicating Factors

In both Figs. 2 and 3, there is an entry which simply says unique materials problems. These are materials behavior problems which are generally unique to a given situation and affect the choice of materials for the design. Nearly all alloys in use today have inhomogeneities and defects in their metallurgical structures. Effort is made to detect defects, but little has been done to assess their effects on creep and rupture behavior. Thus the need for presenting data in statistical terms, since properties of specification materials can vary considerably from lot to lot and even within a lot. Other important considerations are problems of materials instabilities which complicate the life prediction problem, for example, environmental effects, metallurgical instabilities and cracking, and the effects of deformation:

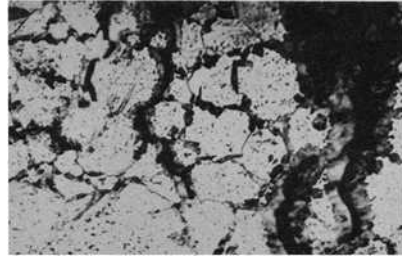
Surface and Environmental Effects—The environment in which alloys operate influence their properties. This ranges from catastrophic failure to simple loss of load carrying capacity due to surface metal removal by corrosive attack[1,2]. Oxidation is the most well known of such phenomena. It may be general, causing either continuous scaling or formation of an adherent protective coating, or preferential, selectively attacking grain boundaries and leading to intercrystalline failures and shortened life. Related phenomena occur when exposed metals absorb carbon, nitrogen, sulfur, or other elements. Rutherford[3] has shown some examples of failures of components in power, oil, and chemical plants due to these factors (see Fig. 4).

A severe environmental problem known as "hot corrosion" occurs in the high temperature operation of gas turbines and jet engines. Fuels used in these engines contain small amounts of sulfur and sodium in the combustion products which apparently are responsible for the attack. "Hot corrosion" can be the limiting feature in alloy design. Figure 5 from Sims[4] illustrates the manner of occurrence and propagation of this attack.

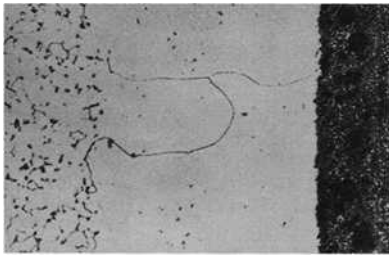
There is an effect of test environment on the creep-rupture properties of metals. This can create a problem for the designer, since most data are taken in the laboratory where air is the normal environment and specimens are small, that is surface to volume ratio is higher than that in actual service components. Comprehensive work was done in this area by Shahinian and Achter[5,6] from 1956 to 1962. Using Nickel-base alloys they found that at high temperatures and low stresses, creep specimens are stronger in air than in vacuum, but at low temperatures and high



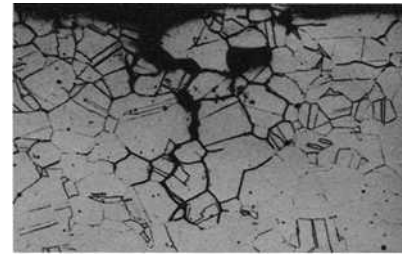
(a) Oxidation



(b) Carburization



(c) Decarburization



(d) Sulfidation

FIG. 4—Metallurgical consequences of environmental reactions.

stresses there is a reversal. Time of test also affected these results. Shephard and Donachie[7] report no environmental effects over a wide range of test variables for oxidation resistant steels.

Finnie and Heller[8] have tried to show how such effects may be accounted for in design procedure, but in general the engineer has relied on the metallurgist to develop alloys which resist particular environmental conditions. The durability of most high temperature equipment attests to the success of this practice, but the engineer must still be aware of environmental hazards.

Metallurgical Instability and Cracking—Metallurgy construes the chain of events: composition $\rightarrow$ structure $\rightarrow$ properties to be that metallurgical structure determines properties and in turn is dependent on composition and processing of the alloy. Unfortunately, alloys are subject to dynamic microstructural changes during elevated temperature exposure. Such changes are time and temperature dependent and are influenced by stress or strain or both. Such changes can induce simple loss of strength or catastrophic failure. This is important to the designer, and he will allow for them if they are recognized. However, most design times far exceed the duration of test data available and the structural changes may not be evident. It is a fact that many such events depend on subtle effects in the process stages of making alloys. Thus a burden is placed on accept-

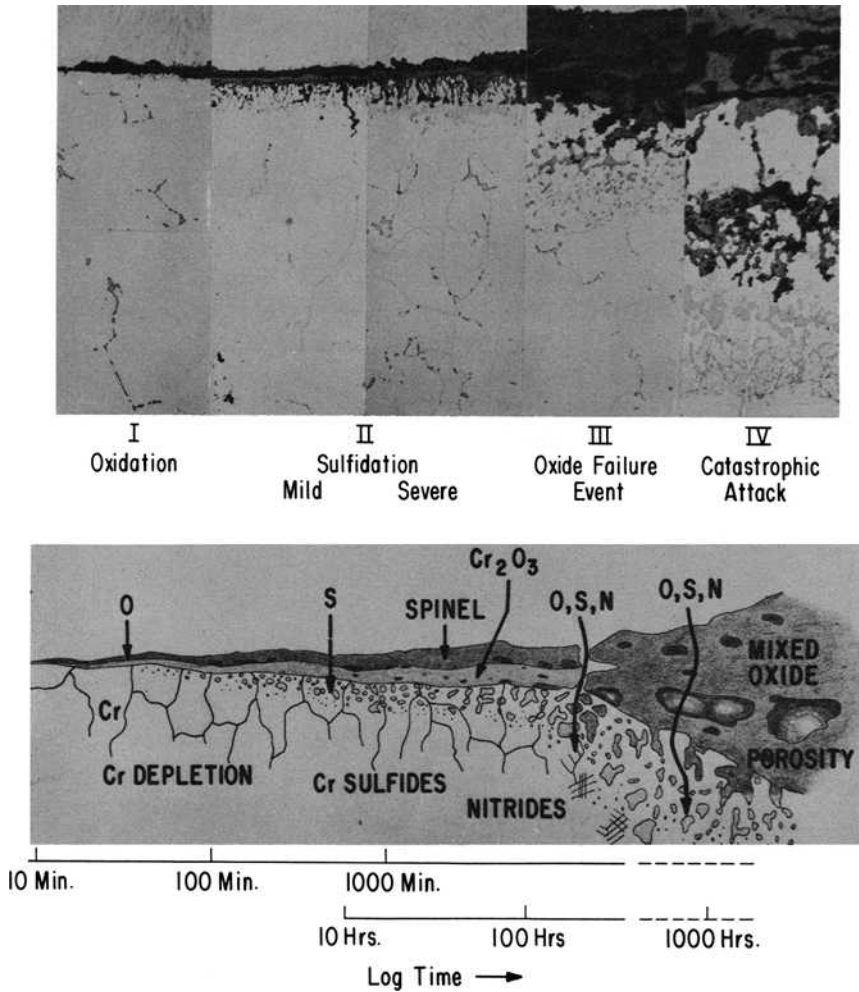


FIG. 5—Steps in the hot corrosion attack of a high chromium Ni-base alloy. Assume diesel fuels marine conditions, at about 1600 F. (Magnification of 300).

ance procedures to qualify materials and on research to identify the mechanism of such changes.

Progress has been made in classifying the microstructural components responsible both for strengthening alloys and for their degradation during exposure. Several excellent reviews in this area are available[4,9-15]. For illustration, Fig. 6 and 7 show the microstructural changes in a low alloy steel after long creep exposure leading to softening and loss of strength and the changes that occur in a nickel-base superalloy leading to embrittlement. Note how the alloy matrix and the dispersed second phases,

Brinell Hardness Number	Notch Bend Strength PSI	Notch Bend Angle O	Post Exposure		
			Rupt Str Hours	Rupt Elong %	Rupt R.A. %

Low Alloy
Cr-Mo-V
Steel

Original Structure

230	210,000	0° 42'	973	19.4	82
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900 F Exposure for 48,063 Hours

224	217,000	1° 56'	313	20.7	63
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1100 F Exposure for 32,785 Hours

168	182,800	7° 12'	4.8	31.7	69
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FIG. 6—Influence of creep exposure on microstructure and properties of low alloy Cr-Mo-V steel.

which are responsible for both the initial strength and the subsequent loss of properties, change with exposure temperature and stress.

Embrittling under creep loading is usually shown by a loss in ductility or toughness (Fig. 7) and involves intergranular fracture where phase precipitation plays a major role. Metallurgists try to prevent the occurrence of such reactions or at least postpone them as long as possible. Unfortunately, such behavior is nonquantifiable and is a deterrent to accurate life predictions.

The instability of most concern to the engineer is the initiation and propagation of the grain boundary crack. In creep there is a contribution to strain from grain boundary sliding which increases with temperature

Magnification = 18,000 Diameters



Time at 1600°F	0	30 Minutes	1000 Hours
Equivalent Time at 1300°F	0	1000 Hours	285 Years
Hardness (Brinell Number)	202	376	350
Toughness (Ft-Lbs)	126	29	2

FIG. 7—Influence of exposure on microstructure and properties on Ni-base René 41 alloy.

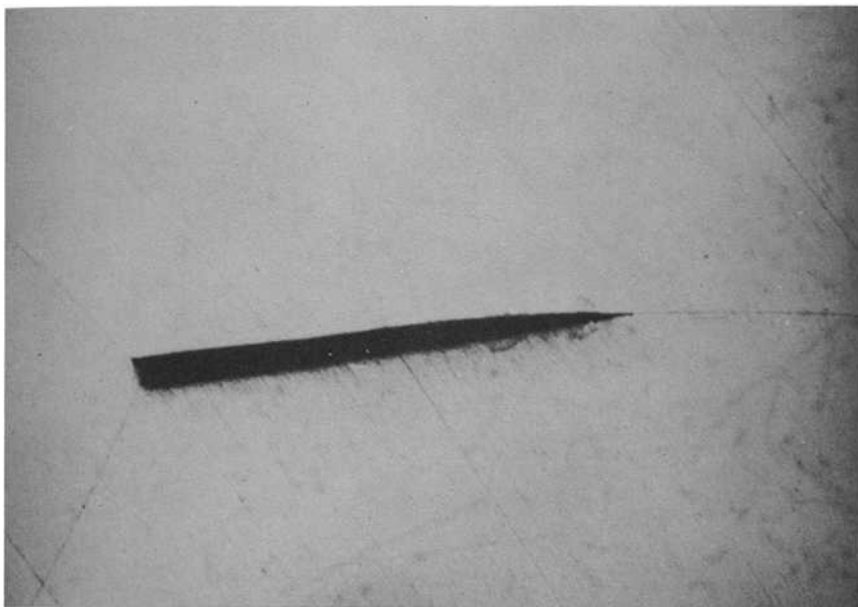


FIG. 8—Triple junction cracking in Al-20 percent Zn alloy at 260 C and 6000 psi (magnification of 150).

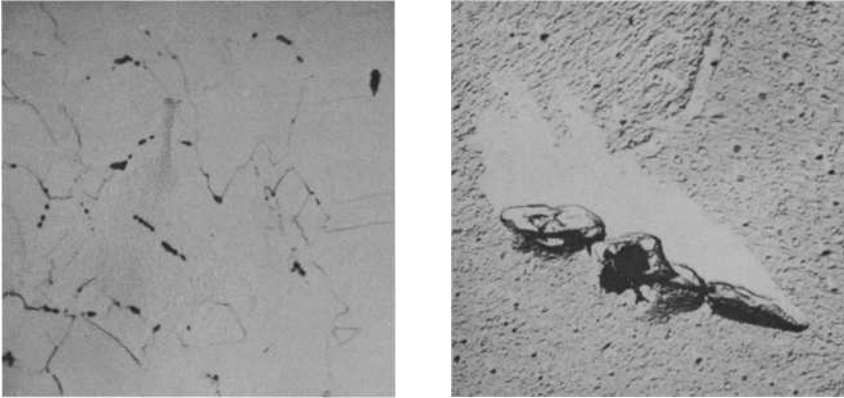


FIG. 9—Well-developed cavities showing some coalescence after creep testing.

and lowering of strain rate. To maintain cohesion between grains under these conditions it is necessary to relax local stress concentrations by plastic strain within the grains. Cracks initiate and grow through grain boundaries when the material is unable to comply with this need. The most likely areas of stress concentration are the particles lying in the plane of the boundary, at the junction of three or more grains, or at mechanical notches due to imposed stress state.

The relation between cracking, continuous change of metallurgical structure, and the imposed stress and strain state during creep exposure is very complicated. Papers dealing with the behavior of grain boundaries and cracking during creep are available[16–21]. The state-of-the art has identified modes of creep cracking, for example, the “wedge crack” (Fig. 8) and the “cavity” (Fig. 9) and has speculated on their mode of formation (Fig. 10) and growth. Figures 11 and 12 show the mode of notch cracking as a function of a low alloy steel’s capacity for deformation. Note the “tearing” or shear mode of cracking at the notch root, the relatively wide and oxidized crack which has propagated transgranularly

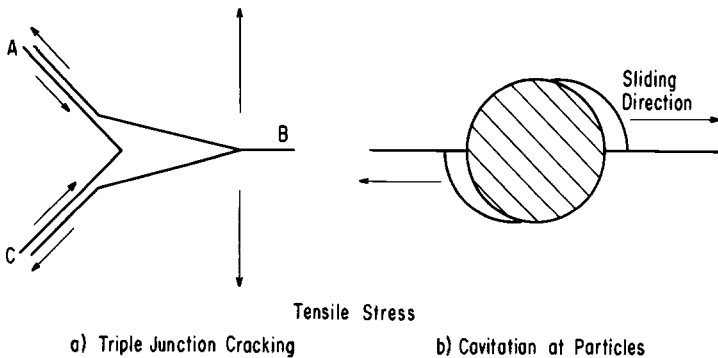


FIG. 10—Grain boundary crack initiation mechanisms.



FIG. 11—Transverse cross section of nearly fractured ductile low alloy steel notch specimen.

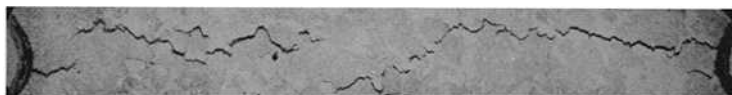


FIG. 12—Transverse cross section of nearly fractured low ductility low alloy steel notch specimen.

for the ductile steel of Fig. 11 and contrast it with Fig. 12. Here the notch root is hardly deformed and cracking has initiated and propagated intergranularly. The cracks are narrow and discontinuous indicating a nonhomogeneous propagation. This latter case is to be avoided since it can cause shortened life and lead to catastrophic failure.

Life prediction is complicated by our lack of understanding of how crack behavior is related to creep test variables.

Creep Deformation—Literature exists on the atomistic[21,22] and microstructural[23] aspects of creep and on the theory and empiricism which have been used to describe creep behavior analytically[24,25]. Some of this behavior is important to our ability to make accurate predictions. In Fig. 13 is shown the components of total strain which are important in creep[26]. Anelastic or recoverable strain as well as the combination of both matrix and grain boundary components in plastic creep strain has been added to elastic strain in the consideration of this problem.

Most attempts at predicting response to changing test conditions have used the "equation of state," that is, total strain is a state variable according to the Ludwig hypothesis[27] where $f(\sigma, \epsilon, T, \dot{\epsilon}) = 0$. At low temperatures where little recovery occurs or at high temperatures where

Strain Component	Time Dependence	Stress Dependence	Recoverability
ϵ_e	Independent	Linear	Recoverable
ϵ_a	Dependent	Linear	Recoverable in Time
$\epsilon_p = \epsilon_g + \epsilon_{gb}$	Dependent	Non Linear	Not Recovered in Time

FIG. 13—Components of total strain.

recovery is rapid, the metallurgical structure is not significantly history dependent and the equation of state is reasonable. In the practical situation the recoverable and grain boundary components are important and invalidate the state equation. Thus, a deficiency in predictive ability at useful temperatures is the inability to account for prior history dependence of structure.

There is evidence that cracks or cavities are initiated continuously during the creep process[16] and may be present early in life. Apparently both time of exposure and total strain influence damage in terms of the cavity volume[19,28]. Furthermore, significant damage is not predictable at any particular stage of creep, nor is the damage at failure constant, that is, damage depends on test conditions[29]. This has engineering importance because the widely used linear damage rule implies otherwise.

Past work done in creep damage has been summarized by Voorhees and Freeman[30]. The importance of the damage problem indicates the need for more study.

Predictive Methods

Current predictive methods involve two problems: first, is the description of material behavior and secondly, the difficult one of extrapolating that behavior. The extreme approaches to this problem lie in the fields of metals physics on the one hand and applied mechanics on the other. Much work has been done in describing the physical mechanisms of creep and rupture but the theories constructed to date are limited to the simplest of materials and conditions of exposure and are not of much practical use. Garafalo[22] has summarized the state-of-the-art in this field. In the applied mechanics view, the materials are seen as a continuum that can be modeled by derived mathematical expressions. However, fundamental laws of plasticity are not rigorously applicable at elevated temperatures and what little is known of metallurgical mechanisms is not applied. In the case of rupture, either the initiation and propagation of cracks is not considered or they are assumed to be independent processes. Thus again the immediate needs of the designer cannot be totally satisfied. Hult[31] and Rabotnov[32] have summarized this area of work.

Today the engineer relies heavily on empiricism. The empirical approach requires taking large quantities of data and its reduction to useful diagrams or analytical expressions which satisfy the needs. Some important empirical predictive methods are concerned with extrapolation of creep and rupture data, nonsteady application of loads and temperatures, complex states of stress, and the influence of notches.

Extrapolation of Creep and Rupture Data—Basically the methods to be discussed apply to uniaxial, constant load, isothermal creep, and rupture data. Figure 14 illustrates some of the ways in which the exposure parameters strain, stress, time, and temperature can be presented for sub-

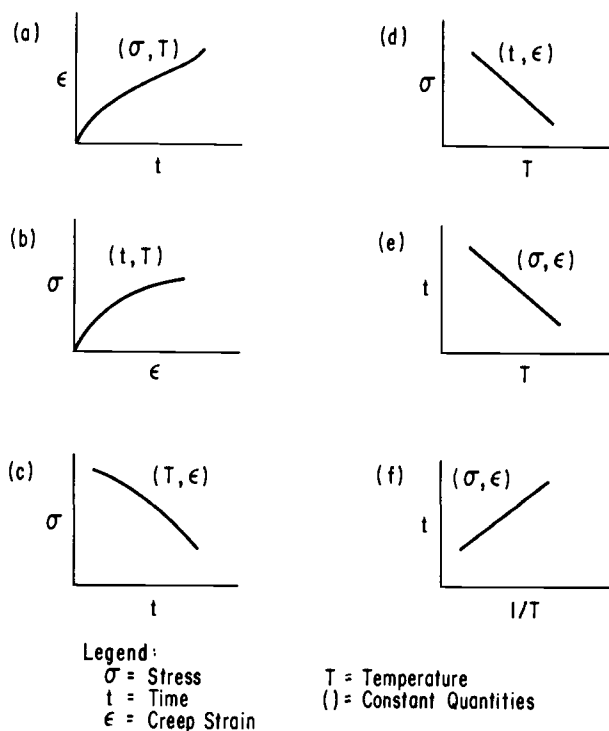


FIG. 14—Basic methods of creep and creep-rupture data presentation.

sequent extrapolation. Figure 14(a) and (b) are isothermal and isochronous creep plots and are used in predicting creep. Figure 14(c) through (f) are common ways to represent both creep and rupture data for extrapolative purposes. The scales used are either linear, logarithmic, or semilogarithmic. Obviously if some law involving all four variables were available any of the curves shown could be extrapolated directly. Currently no general law exists and other means are used.

The obvious way for extrapolation of either creep or rupture data is by direct extension of the curve. This is not an outmoded method. The curve of Fig. 14(c) in logarithmic coordinates is commonly manually extended over several log cycles in time, usually to 10^5 h. The method is preferred by code bodies when dealing with masses of data with scatter. For single sets of rupture data more sophistication can be used in this method as shown by Grant and Bucklin[33], Brozzo[34], Bennewitz[35], and Graham and Walles[36]. These investigators show relationships describing the curvature which can occur over a spectrum of temperatures in plots of the Fig. 14(c) type. Extrapolation then is in accordance with the relationship derived from the existing data. All of these methods require extensive data to describe the instability relationships. Also, metallurgical instabilities can be a function of certain ranges of exposure

parameters and not amenable to such techniques. Here nothing short of the actual data will describe the trends.

A series of curves such as is shown in Fig. 14 (a) can be reduced to a single curve providing for interpolation and extrapolation. Shephard[37] reduced sets of isothermal creep curves to a single curve by multiplying the time scale by a constant, and Dorn[38] tried correlating strain with a temperature modified time using isostress creep curves. Neither method met with great success although the later evolved into a parametric technique to be discussed shortly.

Rabotnov[39] considered a series of isochronous curves as shown in Fig. 14(b). He related total strain to stress and time at constant temperature by an open form equation:

$$\Phi(\epsilon) = \sigma (1 + at^b) \quad (1)$$

Goldhoff[40], using this equation, found the constant b to have a value of about 0.3. In this form a and the quantity $\Phi(\epsilon)$ are described as functions of strain, which allows extrapolation of the curves in time. This form is useful since engineers prefer working with stress-strain data.

Probably the most used predictive technique in the creep-rupture area is the time-temperature parameter which compensates time with temperature in a way that allows prediction of long-time residual strength, or minimum creep rate. The form of these parameters is shown in Fig. 14(e) and (f). Literature reviews[41–44] are available, the most recent of which is Manson's[45]. The time-temperature parameter uses increased temperature for shortened time of test, a concept much used and accepted in treatment of simple chemical reactions. Although the behavior of alloys at elevated temperatures is complex, reaction rate treatment (the Arrhenius equation) at least approximates the behavior in that increased temperature of loading does shorten alloy life. In Fig. 15 a simple treatment of rate theory is given. Some of the assumptions are: (a) substitution of ϵ/t for r suggests an average creep rate whereas the rate changes throughout the life of a test, (b) for the rupture correlation e is assumed constant at failure which is not true, (c) the pre-exponential term and the activation energy are treated as functions of stress and temperature rather than as constant. This is the form that Grounes[46] derives in a more sophisticated treatment. His displays a $[P(T, \sigma) - Q(T, \sigma)]$ matrix from which is derived some thirty time-temperature parameters already proposed. Referring to Eq 4 in Fig. 15 and letting P be a constant and Q a function of stress leads to

$$Q(\sigma) = T(\log t + C) \quad (2)$$

which is the Larson-Miller parameter. In a similar manner we can arrive at other basic forms in common use today as shown in Fig. 16. References 41–46 supply complete background on each of these. These particular

$$1. \quad \tau = A(\sigma, T) e^{-\frac{Q(\sigma, T)}{RT}}$$

$$2. \quad \dot{\epsilon} = A(\sigma, T) e^{-\frac{Q(\sigma, T)}{RT}}$$

$$3. \quad \text{For } \epsilon = \text{Constant}$$

$$t = A(\sigma, T) e^{-\frac{Q(\sigma, T)}{RT}}$$

$$4. \quad 0 = \ln t + P(\sigma, T) - \frac{Q(\sigma, T)}{RT}$$

$$t = \text{Time} \quad \sigma = \text{Stress}$$

$$T = \text{Temperature} \quad \epsilon = \text{Strain}$$

FIG. 15—Derivation of basic time-temperature parameter form from rate equation.

forms are useful because they can be expressed in linear form and plotted as shown in Fig. 14(e) and (f) and in detail in Fig. 16. The implications are seen from Fig. 16 where the iso-stress plots are linear and show either convergence or parallelism of the lines. The result of correlating the data is a master curve as shown in Fig. 17(b). Any parameter number represents various combinations of time and temperature so that long-time residual stress can be extrapolated. One objection to this approach is that the analytical molds into which the various methods force the data are not "universal," and very few sets of data follow the forms shown in Fig. 16. There is an "averaging" process accompanying the "optimized" constant(s) and the shape of iso-stress curves obtained is nonlinear. Figure 17 shows isothermal stress/time-to-rupture data for a superalloy in the upper portion and as a Larson-Miller master curve in the lower portion. The deviation from the master curve correlates with the steep decline in the stress-rupture curve at 1500 F. This correlates with the formation of sigma phase in the microstructure under the exposure conditions. Such metallurgical instabilities cannot be handled by existing parameter techniques.

There is need for a universal method which encompasses all of the common forms and predicts better than those examined to date. Manson[45] has proposed a "minimum commitment-station function" approach to the problem of fitting and extrapolating creep-rupture data. An equation

$$f(\log t) + p(T) = g(\log \sigma) \quad (3)$$

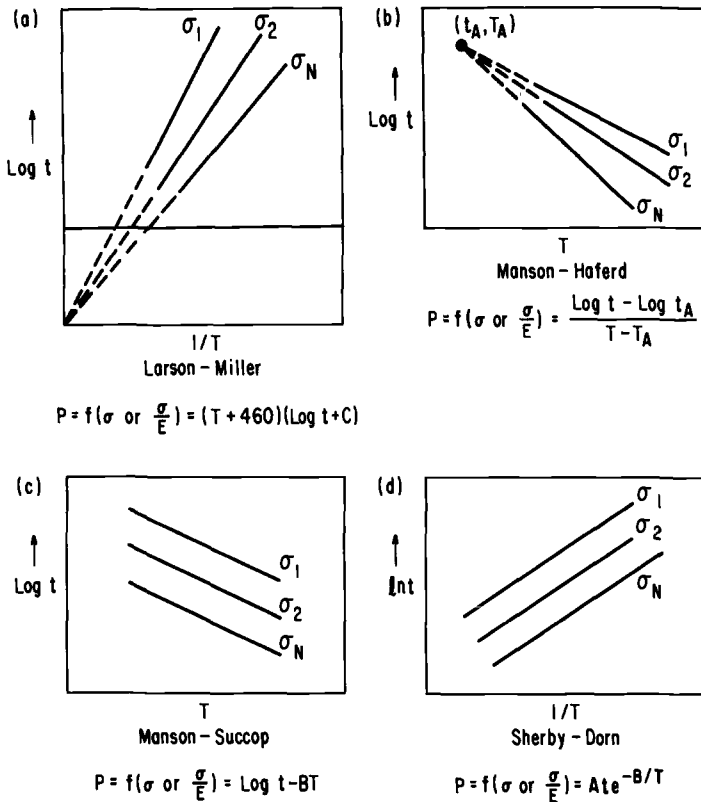
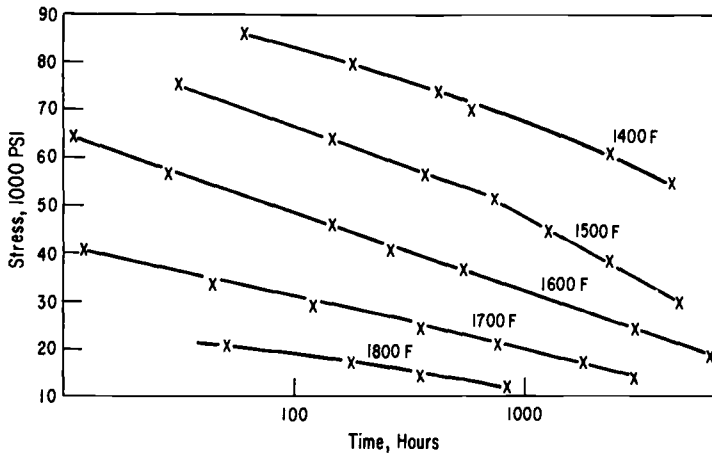


FIG. 16—Time-temperature parameter models.

is assumed, and by rearrangement all common forms can be fitted to this equation. No commitment is made as to the form of these functions but the problem is to find the best form for each function. Station functions are assigned and each data point is translated into a form analogous to Eq 3 by means of interpolation between the station functions. The system of equations is solved numerically for the unknowns after which one of the parameters, (σ, t, T) , can be found if the other two are known. In preliminary tests this method has generally given superior results and Manson is currently working to optimize the technique.

The time-temperature parameter is a desirable tool because it can be used to extrapolate residual creep, creep rate, and rupture strengths. Recently Goldhoff[47] has produced a parametric scheme for extrapolating rupture ductility which is an important material characteristic. It is necessary to find standardized parametric techniques and to outline proper steps for their use in setting allowable stresses and in acceptance testing.

Non-Steady Application of Loads and Temperatures—In practice the conditions of constant stress and constant temperature, seldom prevail. Therefore, methods to predict the behavior of materials under nonsteady



(a) Stress Rupture Curves

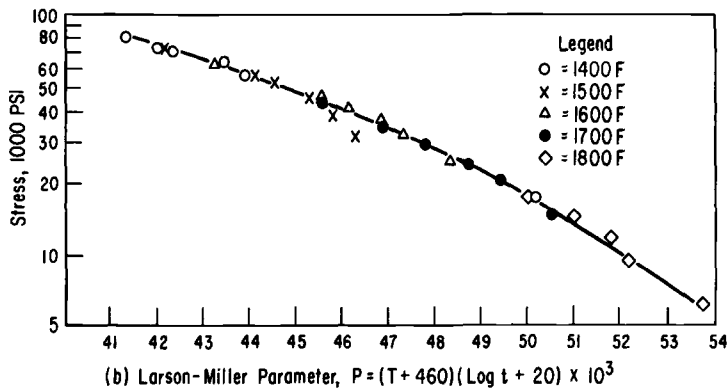
(b) Larson-Miller Parameter, $P = (T + 460)(\log t + 20) \times 10^3$

FIG. 17—Stress rupture and parameter curves for astroloy.

conditions are needed. This means finding a suitable criterion for the accumulation of damage under creep conditions. As indicated, a major deterrent is our inability to assess the effects of prior history on metallurgical structure. Because of this simple approximations are made to account for the dependence of the creep strain or strain rate on prior history. One of these results in a "strain hardening" rule which assumes that accumulated plastic strain accounts for the unknown structure effects while another results in a "time hardening" rule in which it is assumed that elapsed time accounts for the structural effects. These two rules are illustrated in Fig. 18. Finnie[48] has reviewed analytical expression of these rules and discussed them with regard to stress analysis during creep. Varying accuracy is reported with these methods depending on the material used and the duration of the experiment[42].

For predicting rupture under nonsteady conditions the most commonly used criterion is the so called "life fraction" or "linear cumulative damage" rule due to Robinson[49,50]. This is normally expressed as

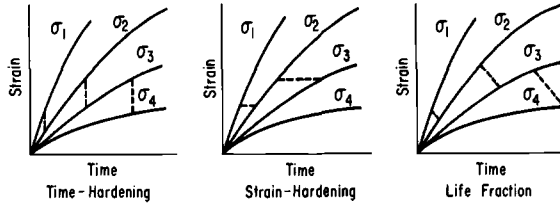


FIG. 18—Schematic curves illustrating proposed creep rules.

$$\sum_i \frac{\Delta t_i}{t_{R_i}} = 1 \quad (4)$$

where t_i is the time of exposure at a given stress and temperature and t_{R_i} is the time to rupture at the same stress and temperature. The rule is shown graphically in Fig. 18. A number of studies give support to this rule[30,51-53], but lacking a sound theoretical basis, it fails in many cases. Goldhoff and Woodford[29] have presented work in justification of nonlinear damage behavior. Randall[51] also suggested limits to this rule. Efforts to improve on the life fraction rule were due to Liberman[54], Oding and Burdukskii[55,56], Grounes [57], and Freeman and Voorhees[30]. Liberman considered a strain fraction rule

$$\sum_i \frac{\Delta \epsilon_i}{\epsilon_{R_i}} = 1 \quad (5)$$

substituting strain for time while Oding and Burdukskii arrived at

$$\sum_i \left(\frac{\Delta t_i}{t_{R_i}} \right)^M = 1 \quad (6)$$

based on a treatment of void initiation and growth. Grounes derived the Robinson rule from generalized rate theory and for design purposes he suggests

$$\sum_i \frac{\Delta t_i}{t_{R_i}} \leq K \quad (7)$$

where K is a constant based on how the creep data comply with certain assumptions. Freeman and Voorhees noted that Liberman's rule was preferable in some cases but in others was inferior to the life fraction rule. This led them to postulate a mixed criterion

$$\sum_i \left[\frac{\Delta t_i}{t_{R_i}} \cdot \frac{\Delta \epsilon_i}{\epsilon_{R_i}} \right]^{1/2} = 1 \quad (8)$$

which they claimed to be more consistent than either of the simple criteria alone.

Day and Cummins[58] studied nonsteady stress and temperature separately and in combination. They conclude that none of the descriptions of

cumulative damage are consistently accurate when applied to their data. The out-of-phase combined stress and temperature changes led to extreme reductions in life and were totally unpredictable. Nonetheless it appears that there is justification for a mixed criterion for cumulative creep damage[20,29,59]. Ata and Finnie[59] suggest that a general rule should contain both time fraction and strain fraction as special cases, apply to increasing and decreasing stresses, and apply to varying multiaxial stresses.

Complex States of Stress—We have been dealing with predictive techniques whose applicability is tested with uniaxial data. Now we must turn to the problem of the multiaxial stress state. It is necessary to be able to correlate uniaxial and multiaxial creep and rupture behavior. Once this is accomplished the same predictive rules should be applicable. However, the determination of criteria for multiaxial creep and rupture is a complicated problem even though generalizations do exist[26]. As Finnie[8,48] points out, the Odquist equations[60] based on ordinary plasticity have received the most use in defining steady state creep while for the case of general creep Bailey[61], and Johnson[62] have derived equations for time hardening and strain hardening. There are copious bibliographies[26,48,63] that deal with analyses for general creep behavior of simple geometries but it is difficult to generalize any of these to practical situations.

For the complex rupture process involving crack initiation and growth, no totally accepted method of fracture prediction under multiaxial creep conditions exists[26]. The most detailed work in this area is due to Johnson[64] and his colleagues. To summarize his findings on rupture, the maximum tensile stress was the best criterion for materials that developed extensive internal cracks prior to failure whereas effective stress was the best criterion for materials which showed little cracking except at the site of final failure. These findings cannot be generalized because they represent two extreme cases. Most engineering alloys are likely to fall in a category somewhere between these limits. A number of studies[65–67] have suggested a mixed criteria, but Johnson and Khan[66] favored the appropriate single criteria. Ata and Finnie[59] reviewed this field and suggested that a rupture life criterion should be written

$$t_R = f(\bar{\sigma}, \sigma_{MAX}) \quad (9)$$

where crack initiation is a function of σ and crack propagation is a function of σ_{MAX} . They presented work on tubes where they expressed the rupture criteria as

$$t_R = Ae - (b_1 \bar{\sigma} + b_2 \sigma_{MAX}) \quad (10)$$

where A , b , and b_2 are constants. They claim evidence to show that $\bar{\sigma}$ does govern crack initiation.

For nonsteady exposure conditions under multiaxial stress time fraction methods require no modification but the strain fraction method requires a combined form of strain. Johnson[68] reported on complex stress tests of aluminum and copper finding the former to be correlated by life fraction whereas the latter fit the strain fraction rule best using effective strain. Ata and Finnie[59] reports test on copper tubes with constant hoop stress and varying axial stress. They propose the following rule designed to fit their criteria

$$d_i = K \frac{t_i}{t_{R_i}} + (1 - K) \frac{\bar{\epsilon}_i}{\bar{\epsilon}_{R_i}} \quad (11)$$

where d is the damage in the time interval t_i during which the effective strain is ϵ_i . K is a constant between zero and one indicating the ability of the material to resist cracking during creep. They also suggest

$$K = \frac{b_1}{b_1 + b_2} \quad (12)$$

where b_1 and b_2 are taken from Eq 10. Better agreement is found with this method than with the geometric mean.

The Influence of Notches—The problem of the notch, inherently a part of most structural components, involves several difficulties. The notch provides a stress and strain concentration on loading. At elevated temperatures the initial stress distribution is elastic-plastic. Relaxation and stress redistribution occur with time, and strain continues to accumulate. With the introduction of finite element computational techniques, a plastic creep analysis to define the stress and strain distribution with time seems hopeful. But even then a cumulative damage criterion would be needed to determine the life. Also, such computation is costly and the engineer will look for approximate methods of predicting notch life data from simpler tests. Voorhees and Freeman[69] actually tried to calculate lives of notched rupture bars using yield, relaxation, and creep data from smooth bar tests and the approaches of classical plasticity theory. They were, however, unable to get realistic results, presumably because of the many simplifying assumptions which had to be made.

The importance of the problem has led to testing of notched bars in uniaxial tension in the laboratory. Since this test is unique in that it can either be weaker or stronger than a comparably loaded smooth bar, much effort is devoted to determining the conditions under which notch strengthening can be maintained. The usefulness of a material is frequently judged in this way and materials are developed based on this criterion. Investigations have examined effects ranging over the duration of exposure, metallurgical structure, size of test bar, geometry of notches, and preparation of the test bar to find how notch bar life will compare to smooth bar life. There are reviews of this literature[11,59,70,71]. The notch introduces a

triaxial stress state and most engineers think of a criterion for rupture in terms of which stress or combination of stresses are responsible for failure. It should be realized, however, that the stress state has an effect on embrittlement, that is, the more severe the triaxiality, the less the ductility. The stress state varies across the notched section and at elevated temperature also changes with time. The ductility will also depend on these factors and in localized areas will affect crack initiation and propagation[11,26].

The engineer then should be interested in whether failure under notched conditions will be stress or strain limited. Goldhoff[72] proposed accounting for effects of stress concentration using limiting strain as a criterion. The approximate method depends on making use of the relations between the elastic stress concentration factor, K_T , and the plastic stress and strain concentration factors, K_σ and K_ϵ as summarized in the generalized Stowell[73] equations due to Hardrath and Ohman[74]. The assumption is that failure occurs when the concentrated strain at the notch root reaches a limiting value. The concept is questionable, since "limiting strain" is hard to define. The uniform strain obtained in smooth creep bars under varying test conditions was chosen as a criterion but is admittedly a poor one. A definition based on crack initiation as a function of strain would be better, but that quantity may depend on the notch configuration.

The point is that some laboratory tests ought to be directed toward defining materials behavior under conditions more nearly like those suggested by real problems and trying to develop useful engineering concepts from the results. Even if the engineer uses computational techniques these results will have to be checked against experimental facts. Crack initiation and growth features of the rupture process have suggested approaches similar to those used in fracture mechanics to several authors interested in elevated temperature behavior[75-78].

Design Aspects and Test Needs—The mechanical designer may be interested in predicting creep deformation, rupture life, or may even consider cracking and the growth of cracks as limiting features in his design. What tools does he have available? To predict deformation he may use an average stress analysis with no particular model in mind and obtain a gross stress picture; an elastic analysis with some particular model in mind but which is not correct in the creep range; a steady-state creep analysis giving a limit type of solution valid only under special circumstances; or a transient creep analysis which includes elastic, plastic, transient, and steady-state creep behavior. The later is only now becoming feasible as numerical solutions can be obtained using computers. For rupture he relies almost entirely on laboratory data and their extrapolation to arrive at allowable stresses. Current practice is concerned with rupture and only recently has crack initiation and growth been considered. Design pro-

cedures are rather elementary, and the difficulty lies in gathering and interpreting data.

In the introduction, the example of the turbine valve was used as a component designed on the basis of creep and rupture behavior. The intersection of the entry part with the main body of the valve is a critical region, since it represents a notch configuration which could be a point of failure in service if improperly designed. The question is asked if stresses and strains can be predicted analytically for this case. First an instrumented hydrostatic test was done at room temperature to obtain the elastic stress at the intersection[79]. A two-dimensional model was used to obtain equations for principal stresses which when combined into an effective stress did reasonably predict the measured stress. Next, a pressurized cylinder was tested to failure at temperature under realistic pressures. The equations developed were applied to this geometry to predict effective stresses at the inner and outer surfaces of the cylinder assuming relaxation to an average hoop stress at temperature. The effective stress and the stress concentration for the cylinder were calculated. These data were then used in conjunction with the modified Stowell analysis referred to previously. This method can generate stress relaxation and strain accumulation behavior for the given notch. For this step the creep and rupture data for the material in question had to be available.

As shown in Fig. 19, the calculated relaxation curve is compared with the actual stress rupture curve, the intersection being the predicted failure time of the cylinder. Similarly, calculated strain accumulation is compared to the curve of actual rupture elongation with time and the inter-

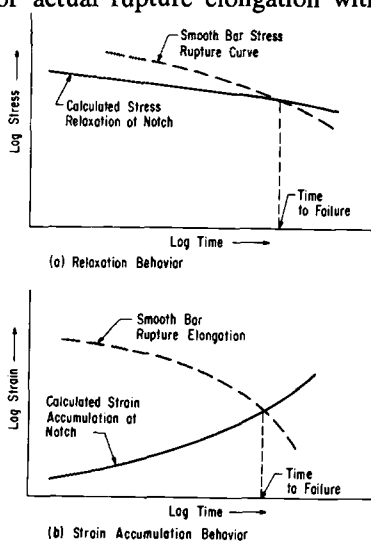


FIG. 19—Predicting failure at stress concentration from smooth bar data and Stowell relations.

section again represents time to failure for the cylinder. Agreement with test was good as the actual value was bracketed by the calculation. Finite element techniques were used to get plastic solutions and they agreed well with the approximate method. So the designer has some assurance that applying the approximate technique to the actual geometries and operating conditions will allow him to predict lives for a variety of designs. He constantly looks for design changes that can reduce stresses and strains and then considers what material characteristics will insure reliable operation, that is, what strength and ductility combination is needed. If the machine is to accumulate numerous stops and starts, he may consider damage accumulation or some aspects of nonsteady loading. Thus, one specific solution to the general problem.

In summary, current design practice trends to realistic analyses in which the modes of material behavior are factored in to the degree in which they are known or can be approximated. The analytic capability has been greatly enhanced by the computer, but this requires experimental backing to verify results. For realistic life prediction, more complex material testing will be needed to calibrate analyses and to explore the fundamental aspect of material behavior.

Fatigue at High Temperature

Mechanical Aspects of High Temperature Fatigue

We subdivide the fatigue problem into two categories, low cycle fatigue and high cycle fatigue. High cycle fatigue can be defined as that situation where the structure is acted on by cyclic strains whose plastic components anywhere in the structure are sufficiently small that an elastic analysis satisfactorily describes the behavior. On this basis, low cycle fatigue can be defined as that situation where the cyclic plastic strains represent a significant fraction of the total cyclic strains in critical portions of the structure. Since there is substantial variation from material to material in the response to cyclic plastic strain depending on the specific values of strength and ductility at the temperature of interest, we cannot select a specific number of cycles as the demarcation between high and low cycle fatigue. Instead, it has been proposed[80] that this distinction be based on the transition fatigue life, defined as that life which a particular material withstands when the elastic and plastic strains are of equal magnitude. The vast difference in life under these conditions, exhibited by three different materials, cast Udimet 500 at 815 C, AISI 304 stainless steel at 650 C and high purity tantalum in argon at 730 C are shown in Fig. 20, where the transition fatigue lives are seen to be 20, 10 000 and 200 000 cycles to failure, respectively. This definition, then gives a rational basis for determining if a particular application is high or low cycle fatigue.

Identification of the failure as low or high cycle fatigue is important in

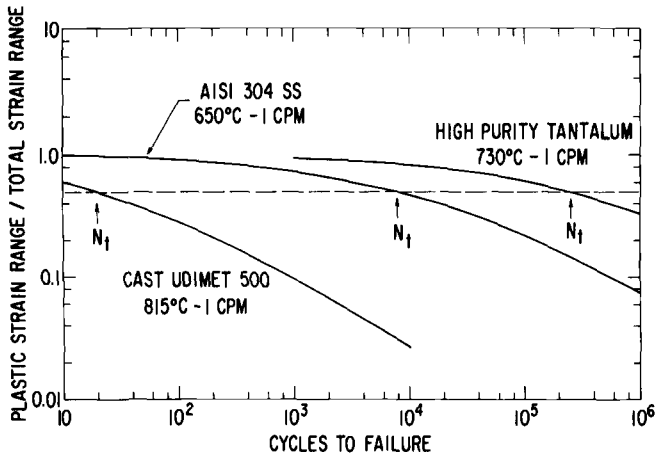


FIG. 20—Ratio of plastic strain range to total strain range versus cycles to failure and transition fatigue life N_t for three materials.

the determination of the analytical method for evaluating the cyclic stresses or strains in the structure. A high cycle fatigue problem immediately defines the analysis as elastic, while a low cycle fatigue problem requires nonlinear analytical procedures for the determination of the cyclic stresses and strains[81]. An alternate approach, which has doubtful reliability in all cases, is to introduce the concept of the pseudostress[82]. Here the assumption is made that the strain distribution from an elastic analysis is valid for nonlinear problems, so that a strain comparison between material behavior and design can be made. The pseudostress is that stress determined by the product of the strain and the elastic modulus of the material.

The determination as to whether one is dealing with a high or low cycle fatigue problem is important also from consideration of the test method to be used in material evaluation. If the problem is a high cycle one, then stress controlled experiments are meaningful, while if the problem is in the low cycle region, only strain controlled experiments have meaning.

Establishment of the transition fatigue life does not set a sharp boundary between high and low cycle fatigue, as seen in Fig. 20. From the choice of analysis (elastic versus plastic) or of the test method (stress or strain control), that life where the plastic strain is, say, only 5 percent of the total strain range would be such a boundary. Thus high cycle fatigue for cast Udimet 500 at 1500 F would be for lives in excess of 3000 cycles, while for AISI 304 stainless steel at 650 C it would be greater than 10^6 cycles. It will be seen later that frequency of cycling must also be considered.

Sources of cyclic stress or strain come from either mechanical or thermal effects. Mechanical effects include cyclic pressure, centrifugal loading, mechanical, and acoustic vibration, etc. Thermal effects arise

from transient thermal stresses developed in reaching steady-state thermal gradients, thermal expansion mismatch in materials, etc. Cyclic stresses or strains of thermal origin are generally accompanied by a cyclic temperature, and this complicates the problem considerably[83]. Often the critical fatigue problem occurs in a notch or fillet and the surrounding material behaves in an elastic fashion. In these cases the fatigue problem can be treated as one involving cyclic strain alone. There are many examples, however, where cyclic strains are produced from thermal or mechanical causes, and, simultaneously, the mean strain changes in some time-dependent way. An example of this is a gas turbine bucket subjected to cyclic plastic strain at the leading and trailing edges due to start-stop operations, and simultaneously to progressive creep elongation associated with the action of centrifugal stresses. This behavior can be simulated in the laboratory for example by the introduction of a straining program in which both cyclic and mean stresses or strains are developed with time[84–89].

Physical Aspects of High Temperature Fatigue

Physically the fatigue process can be characterized as a process of initiation and growth. At low temperature the initiation and early growth processes can often be related to crystallographic factors. Initiation of cracks is often found to occur along annealing twin boundaries or slip bands and are often identified by extrusions. Inclusions, casting defects, or other faults often serve as microcrack sources. Initiation can occur quite early in life (1–2 percent), often at multiple sites. Growth processes involve separate growth of individual sites, followed by crack joining, until a single stable growing crack results. At higher temperatures, grain boundary cracking is most commonly observed, particularly at low frequencies and higher temperatures. Gell et al[90] review the observations for certain nickel base superalloys. McMahon and Coffin[91] have studied cast Udimet 500 subjected to low cycle fatigue at 1500 F and show the initiation to be the result of a localized ridging, Fig. 21. Propagation occurs along grain boundaries accompanied by substantial oxidation, Fig. 22. They propose a stress-oxidation mechanism for crack initiation and propagation, specifically as it relates to grain boundaries. The presence of oxide filled intergranular cracks is a common observation in high temperature fatigue, and this observation will be referred to again in the later discussion.

Modeling the Fatigue Process

Fatigue cracks initiate in regions of localized plastic strain, they grow first within this region and then become stable and grow beyond the plastic zone into an essentially elastic regime, propagating by their own plastic zone developed at the crack tip. These three processes are shown in Fig.

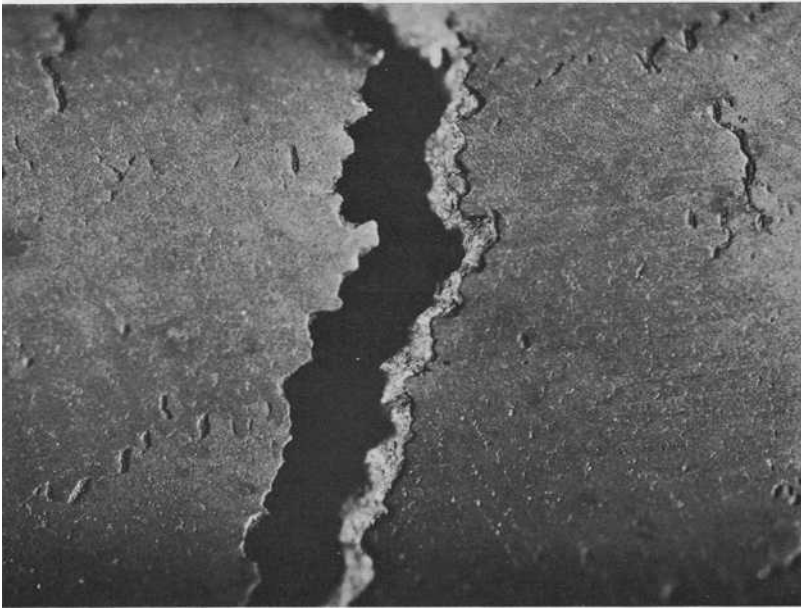


FIG. 21—(a) Surface of specimen after fourteen loading cycles at 815 C showing cracked ridges. (b) Same area as (a) After fracture at 28 cycles showing fracture path running through cracks shown in (a). (Magnification of 22).



FIG. 22—(a) *Profile of crack in specimen A after twenty-five loading cycles located at minimum cross-section. (Magnification of 260).*

23 and for each we can develop phenomenological relationships for purposes of design. Considering first crack initiation, it is assumed that a small test specimen can be located in the critically-strained region and subjected to the same strain history as that in the plastic zone of Fig. 23. This specimen can be considered to be a standard hour-glass shaped low cycle fatigue specimen used in the laboratory. We further assume that failure of this specimen in a laboratory test constitutes crack initiation to an observable size crack in the real structure. By this means data from laboratory tests can be applied to the actual structure. If the plastic zone is indeed in a notch, the equivalence to uniform specimen behavior can be approximated by the use of "Neuber control" as proposed by Topper, Wetzell and Morrow[92]. Failure of the structure can be defined as $N_f = N_i$ although it would be a conservative definition.

A second stage in the modeling of process is to evaluate the crack growth rate as it propagates through the plastic zone. Methods for this



FIG. 22 (Continued)—(b) Detail of tip of crack in (a) showing tendency of oxidation to follow grain boundary. Note Area of depleted γ' at periphery of oride. (Magnification of 780).

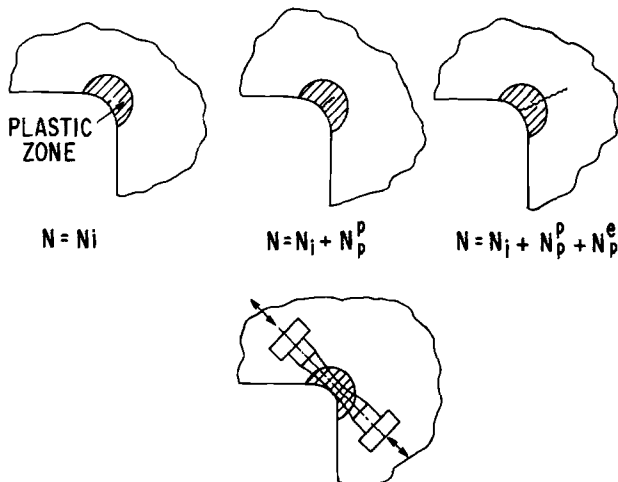


FIG. 23—Stages in fatigue failure of service component and method of modeling.

are not well developed but recent work by Solomon[93] shows great promise. He has shown that the crack propagation law in controlled cyclic plastic strain is

$$\frac{dc}{dN} = \phi c \Delta \epsilon_p^\alpha \nu^{k-1} \quad (13)$$

where

c = crack length

α = a constant, the reciprocal of the exponent of the Coffin-Mason equation, and

k = a frequency exponent (to be considered later).

The size of the plastic zone shown in Fig. 23 depends on the geometry, the material, and the loading. Equation 13 allows the determination of the number of cycles required for the crack to pass through the plastic zone N_p^P . Combining the two procedures failure can now be defined as $N_f = N_i + N_p^P$.

Finally, consideration must be given to elastic crack growth. Although considerable attention has been given to crack growth rates at room temperature, relatively little work has been done at elevated temperature. Reference should be made to the work of Johnson[94], and to Coles and Popp[95]. In the later case, the crack propagation rate of IN-718 was studied at several temperatures and hold times. For a fixed stress intensity range $\hat{K}$, a single valued function of crack growth rate could be related to a Larson-Miller parameter of the form

$$P = (T + 460) (c + \log \tau_h) \quad (14)$$

where τ_h is the hold time in hours per cycle, as shown in Fig. 24.

Use of elastic crack propagation permits consideration of fracture toughness and the critical crack size in the design. This is particularly important in those applications where high strength materials are utilized at high mean stress levels. Thus the quantity N_p^e can be determined, based on fracture toughness considerations, or some other limitation (fluid leakage, vibration, etc.). Finally, then failure can be defined as $N_f = N_i + N_p^P + N_p^e$.

Predictive Approaches

We next consider the state of the art relative to the prediction of life under high temperature fatigue. Approaches which consider cyclic and time dependent effects as separate phenomena will first be discussed. Here the damage fraction of each mode is determined and combined, assuming a linear damage accumulation rule. The early work on creep and fatigue interaction of Lazan's[88] is based on elevated temperature combined cyclic and mean stress experiments carried out at high frequency on S-816. He found that with increasing temperature, increasing amounts of

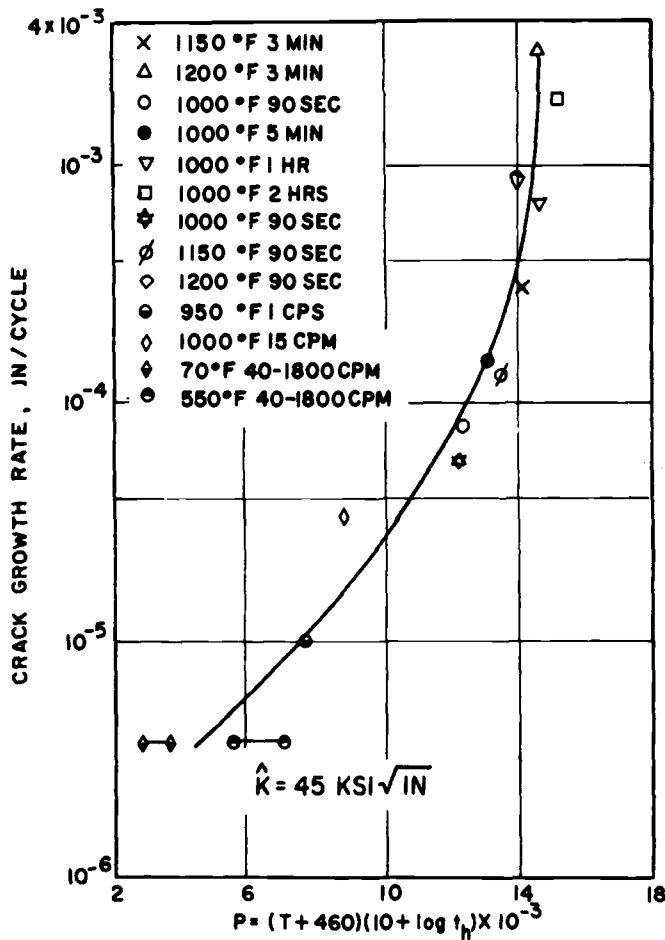


FIG. 24—Parametric fit of crack growth data at constant $\hat{K}$, after Coles and Popp[95].

cyclic stress were required to effect the rupture life caused by the application of a mean stress. Correspondingly these cyclic stresses caused a pronounced reduction in rupture ductility. A model of the behavior was developed, based on a steady-state creep law (power law relation between stress and strain rate) combined with the instantaneous stress to predict the total creep. Further, by using dimensionless modified Goodman diagrams he investigated the applicability of various damage rules to high temperature fatigue-creep behavior. Results are shown in Fig. 25 for S-816 alloy. Taira and Koterazawa[89], following similar lines, refined the analysis by employing a primary creep law. As indicated, earlier designs based on purely stress considerations are difficult to apply to complex structures where strain, rather than stress, is the independent variable.

The creep-fatigue interaction predictive approach is built in part on the background of the work discussed above, coupled with the assump-

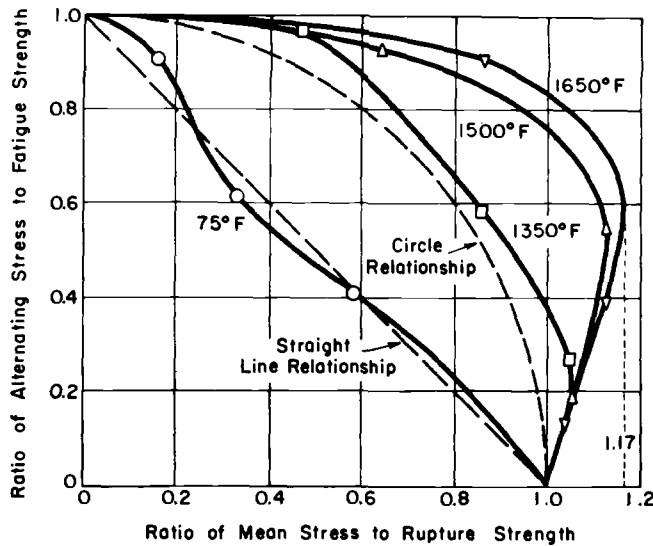


FIG. 25—Unitless stress range diagram for S-816 Alloy at temperatures of 75, 1350, 1500 and 1650 F. Fatigue strength based on 2.16×10^7 cycles and rupture strength on 100 h. (Time required for 2.16×10^7 cycles in test) after Lazan[88].

tion of linear cumulative damage rule proposed by Robinson[50] and later by Taira[96]. This rule can be stated as

$$\phi \text{ fatigue} + \phi \text{ creep} = 1 \text{ (at failure)} \quad (15)$$

Taira and Ohnami[97] considered such an approach in experiments on combined cyclic thermal stress and mean stress on AISI 347 stainless steel, through the utilization of a Goodman diagram representing their results.

Swindeman[98] performed both stress-rupture and square stress wave fatigue tests on Inconel at 1500 F, and found that, using the time under the tension stress, the data agree to within a factor of 2 on a time basis (or a factor 1.2 on stress). Comparison of fatigue under various mean stresses on a Goodman diagram show good correlation except at low mean stress. Evidence was also given, based on strain controlled tests, to show that under combined creep and fatigue, the fractional damage under each condition can be summed as shown in Fig. 26.

Manson, Halford and Spera[99] investigated the elevated temperature low cycle fatigue behavior for two high temperature alloys, AISI 316 stainless steel and L-605 tested over a range of frequencies, strain ranges and temperatures. Observed cyclic lives were compared with lives computed on the basis of a linear creep-fatigue damage rule. Here the pure fatigue life for a given cyclic strain range was determined by the method of universal slopes and the fatigue damage is then computed as a ratio of the number of applied cycles to the pure fatigue life. The pure creep-rupture resistance was evaluated using a cyclic creep-rupture test wherein

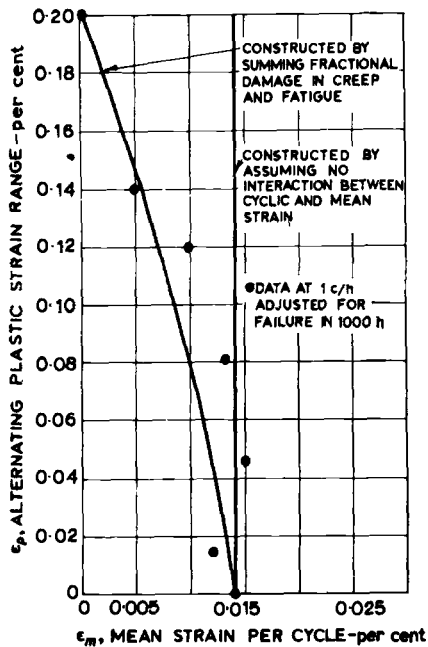


FIG. 26—Combination of cyclic and mean strain to produce failure in 1000 h for Inconel at 1500 F. After Swindeman[98].

the direction of the rupture stress was reversed each time the creep strain reaches a preset tensile or compressive strain limit. Plotting the cyclic rupture stress versus the elapsed rupture time under only the tensile portion of the loading determined the pure creep-rupture curve used in computing creep damage. Creep damage was defined as the ratio of time under a given stress to the time to rupture under the same stress. Summing the creep and fatigue damages and equating the sum to unity establishes the failure criterion in accordance with the linear creep-fatigue damage rule referred to above. Figure 27 shows the method as applied to L-605 at 1180 F at a frequency of 0.167 Hz.

Earlier Spera developed a method of accounting for creep in both thermal and isothermal strain cycling[100], in which time ratios and cycle ratios served as the measure of creep and fatigue damage. The fatigue component was treated as above[99], except that the stresses were measured or determined by detailed computation for a number of experimental situations that served as test cases. The creep damage caused by a given stress acting for a given time required the stipulation of a creep-rupture curve. Here two curves were considered, the conventional monotonic creep-rupture curve for smooth specimens and the curve for sharply notched specimens. These curves provide lower and upper bounds, respectively. It was found that nearly all cases the test data did indeed fall between the computed bounds.

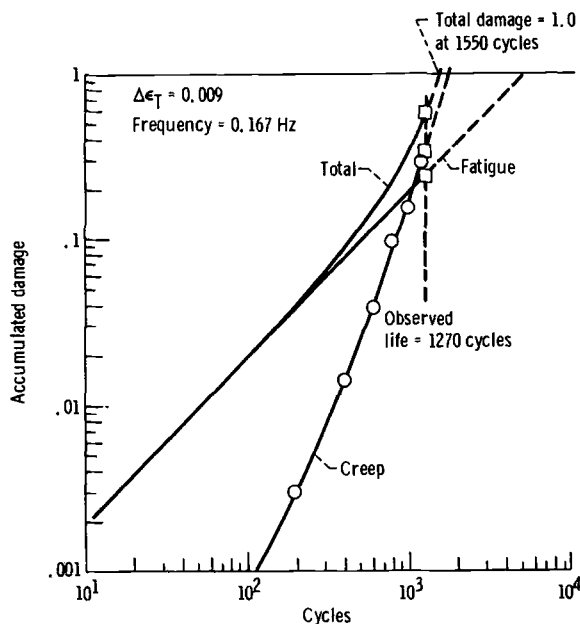


FIG. 27—Progressive increase in creep and fatigue damage during strain cycling of L-605 at 1180 F. After Manson, Halford and Spera[99].

The second grouping of predictive procedures is built, more or less, around the concept that high temperature fatigue can be predicted from the fatigue process alone, provided corrections are introduced to account for the time dependency. The first of these was proposed by Manson and Halford[101] and involves the extension of Manson's method of universal slopes[102] to elevated temperature. A set of criteria is established which determines a lower bound life which is 10 percent of the life predicted in the methods of Ref 102. Some results of this method are shown in Fig. 28. A still lower life, based on the inclusion of the frequency, the slope of the stress-rupture line, the fraction of time the material is subjected to maximum stress in the cycle and certain characteristics of the creep-rupture are included in the method. This later correction is, in reality, an application of Eq 15, and belongs in the predictive methods discussed above.

A second method is that proposed by Timo[103] who develops a family of strain controlled fatigue tests, which, in the low cycle regime, are constructed to be consistent with hold time tests of increasing duration and which in the high cycle regime are related to stress-rupture results. For a curve representing a particular hold period of for example 3 h, the total strain range in the high cycle regime is determined from a rupture strength of the material corresponding to the time of the test, that is the hold period times the number of cycles to failure. A typical curve is shown in Fig. 29.

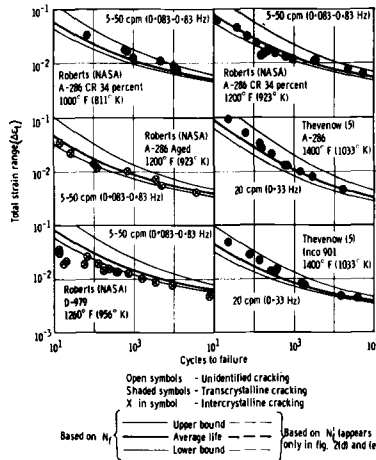


FIG. 28—Comparison of estimated and observed fatigue behavior for Iron-Nickel-Chromium-Molybdenum Alloys, After Manson and Halford[101].

Eckel[104] and, more recently, Coles and Skinner[105] have proposed the use of a frequency-time parameter of the form $\nu^k t$ in expressing high temperature fatigue behavior of various frequencies. Here ν is the frequency, t is the failure time and k is a material constant. They have applied this parameter to data where the total strain range is maintained constant, as shown in Fig. 30.

Conway, Berling and Stentz[106] have proposed a method for predicting hold time results from test results using no hold periods. They show that, for a given total strain range, the time to fracture is always

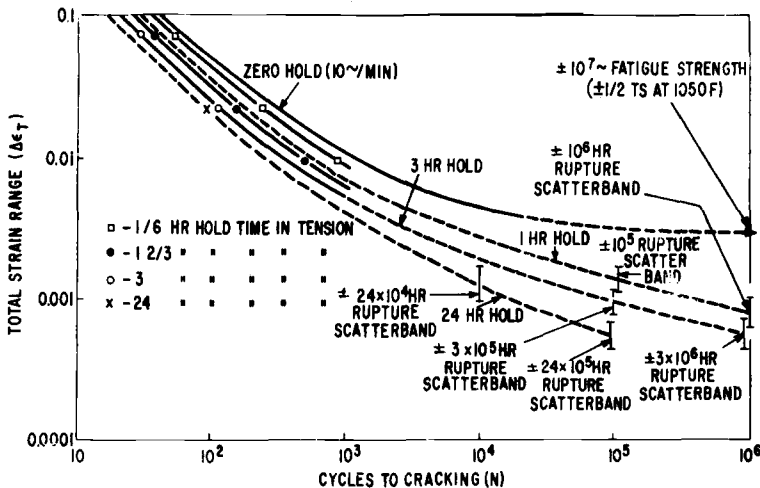


FIG. 29—Corrections for hold-times from Coles and Skinner data, after Timo[103].

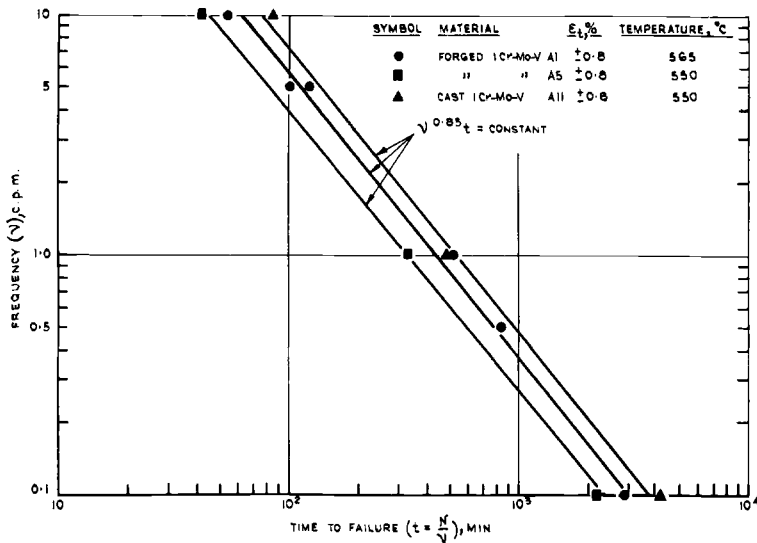


FIG. 30—Influence of frequency on time to failure, After Coles and Skinner[105].

greater when hold periods are involved, even though hold periods reduce the number of cycles to fracture. Further, they observed a linearity on a log-log plot of the time to fracture and the length of the hold period in tension-hold-only tests, from which the time to fracture for any holding time can be calculated. It can be shown in what follows that this procedure is a direct consequence of the high temperature fatigue equations to be described below.

The Development of Fatigue Equations for High Temperature

Models for fatigue life prediction at elevated temperature, based on the separate considerations of creep and fatigue and their eventual summation, are subject to question for at least two reasons. Since the actual fatigue process in a real component is as pictured in Fig. 23, it is unclear how the separate summations of a fatigue process occurring alone at a substantially different frequency, and a creep process acting by bulk deformation and fracture at a fixed load actually contribute to cause the failure. Secondly, the time dependency associated with the predictive requirements are assumed to be the result of creep, a diffusion process, and no consideration is given to the role of the environment in the failure, except as it may influence the separate fatigue and creep-rupture behavior.

A different approach for high temperature fatigue prediction has been developed recently based on the introduction of frequency terms in the various phenomenological relationships commonly in use for fatigue at low temperature[107]. The development of these equations has been described in recent reports[108–111] and the reader is referred to these references for details. The equations are:

$$\Delta\sigma = A' N_f^{-\beta'} \nu^{k'_1} \quad (16)$$

$$\Delta\sigma = A \Delta\epsilon_p^n \nu^{k_1} \quad (17)$$

$$\Delta\epsilon_p = C_2 (N_f \nu^{k-1})^{-\beta} \quad (18)$$

Additionally

$$\Delta\epsilon_e = \frac{\Delta\sigma}{E} \quad (19)$$

and

$$\Delta\epsilon = \Delta\epsilon_e + \Delta\epsilon_p. \quad (20)$$

where

- $\Delta\sigma$ = half-life stress range,
- $\Delta\epsilon_p$ = half-life plastic strain range,
- $\Delta\epsilon_e$ = elastic strain range,
- $\Delta\epsilon$ = total strain range,
- N_f = life
- E = modulus of elasticity, and
- ν = frequency, cpm.

Equation 16 relates stress range to cycles to failure and frequency. Note that when $k'_1 = 0$, Eq 16 becomes the Basquin equation[107]. Equation 17 relates the stress range to the plastic strain range and frequency. When $k_1 = 0$, Eq 17 becomes the well known cyclic stress—plastic strain relationship when the hysteresis loop has stabilized. Finally, Eq 18 relates the plastic strain range to cycles to failure and frequency[108]. When $k = 1$, Eq 18 becomes the Coffin-Manson equation for time-independent low cycle fatigue behavior. A more detailed treatment of the precursor time-independent equations is given by Raske and Morrow[107], and by Landgraft[80].

It should be noted that the three high temperature fatigue equations are not independent. Actually any two will suffice to describe the behavior, and the third can be derived. Thus for a given set of test results there are six coefficients to be determined of the nine defined in Eqs 16–18. The other three can be derived. Choice of the equations selected to calculate the coefficients depends upon the accuracy of the test results. Generally Eq 16 and 17 involve less scatter particularly when materials of low ductility are considered, and plastic strains are difficult to measure[111]. In such cases, test data for Eq 17 may be obtained in a different manner than described above. Here a single specimen can be employed, and subjected to a variety of frequencies and plastic strains and the stress range response measured after the hysteresis loop for a particular pair of frequencies and plastic strain ranges has stabilized.

The fit of the test results with Eqs 16–18 can be examined in a variety of ways. One way is to represent the test results in a manner similar to that at low temperatures where the material is time-independent. Here the cycles to failure are plotted logarithmically as the abscissa, while a

logarithm strain scale is the ordinate, and the elastic, plastic, and total strain ranges are plotted on this scale. At high temperature this method must be modified since a separate curve of strain versus life exists for each frequency. To correct for this, a frequency-modified elastic strain and plastic strain can be found, using Eqs 16 and 18. Experimental quantities are plotted versus the life on Fig. 31 for A-286 at 1100 F and are compared with analytically derived frequency-modified strain-life relationships using coefficients given in Table 1 for this and other materials.

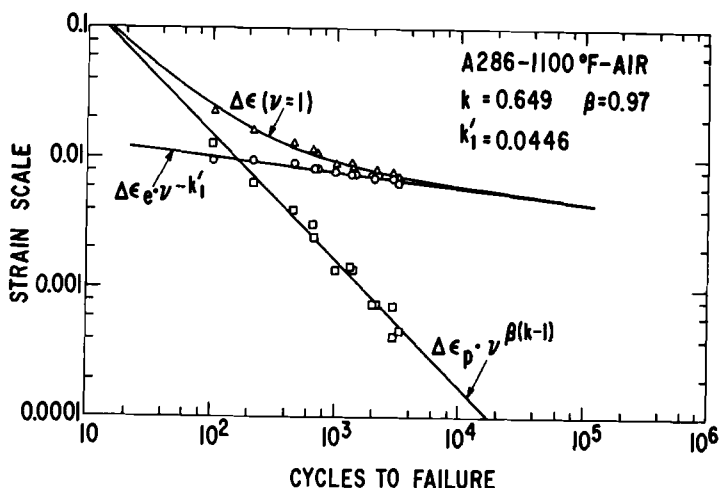


FIG. 31—Frequency-modified elastic and plastic strains for A-286 at 1100 F in air. Total strain is for frequency of 1 cpm only.

Equation 17 can also be compared to the experimental results. By rearranging Eq 17

$$\Delta\sigma\nu^{-k_1} = A(\Delta\epsilon_p)^n \quad (21)$$

The left hand side of this equation is defined as the frequency modified stress range. These frequency modified test values are plotted versus their corresponding plastic strain ranges in Fig. 32 for A-286 at 1100 F and C-1010 steel at 1112 F. The solid line in that figure is the analytical representation of Eq 21. Together they indicate the validity of Eq 17 for the two materials.

Another way of comparing the fatigue equations to the test results is through the use of the total strain range, from Eqs 16–18, or

$$\Delta\epsilon_t = C_2(N_f\nu^{k-1})^{-\beta} + \frac{AC_2^n}{E} N_f^{-\beta n\nu^{k_1+\beta n(1-k)}} \quad (22)$$

By applying the constants of Table 1 for a particular material to Eq 22, a comparison can be made of the derived and experimental total strain range results. This is shown in Fig. 33 for A-286 at 1100 F.

TABLE 1—Coefficients for high temperature fatigue equations for several materials.

Coefficient	A286-595 C		AISI 409		AISI 304		AISI 316		AISI 348		Udimet 500	
	Air	Vacuum	427 C	650 C	816 C	650 C	816 C	650 C	816 C	650 C	816 C	816 C
A'	446 000	690 000	125 000	226 000	68 000	216 500	99 000	152 100	88 400	160 740		
A	427 000	916 000	118 000	214 100	63 260	266 000	94 250	181 900	106 860	520 000		
β'	0.12	0.16	0.052	0.187	0.094	0.143	0.117	0.102	0.131	0.132		
k'_1	0.045	0.12	0.007	0.089	0.071	0.074	0.091	0.038	0.127	0.075		
n	0.192	0.35	0.064	0.258	0.105	0.24	0.151	0.191	0.224	0.185		
k_1	0.0026	0.116	0.0013	0.053	0.053	0.052	0.079	0.015	0.115	0.057		
C_2	1.32	0.37	2.65	1.108	1.72	0.354	1.081	0.382	0.365	0.0736		
k	0.65	0.99	0.889	0.81	0.81	0.86	0.90	0.77	0.92	0.864		
β	0.97	0.44	0.831	0.707	0.87	0.57	0.74	0.527	0.56	0.712		

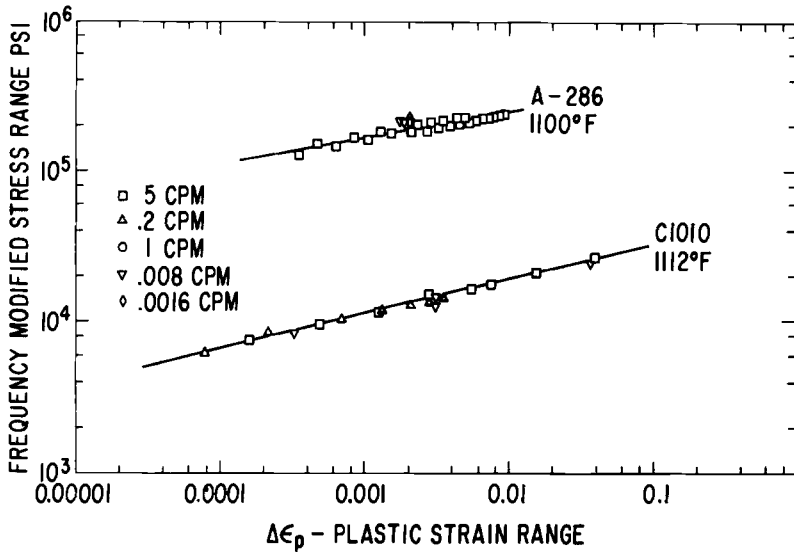


FIG. 32—Frequency-modified stress range versus plastic strain range for A-286 at 1100 F and C-1010 steel at 1112 F.

The derivation of Eq 22 and the material constants of Table 1 come from laboratory tests involving controlled strain limits and cyclic strain with no hold times. Of particular interest is the possible application of this approach to hold time test results. Laboratory hold time tests generally involve a strain-time program shown in Fig. 34b. Here a fully reversed strain of period τ_c' is imposed together with a tensile strain held for a time τ_H , such that the total period is $\tau_c' + \tau_H$.

If one now assumes that wave shape does not influence the fatigue results, hold time results could be predicted from triangular wave shape

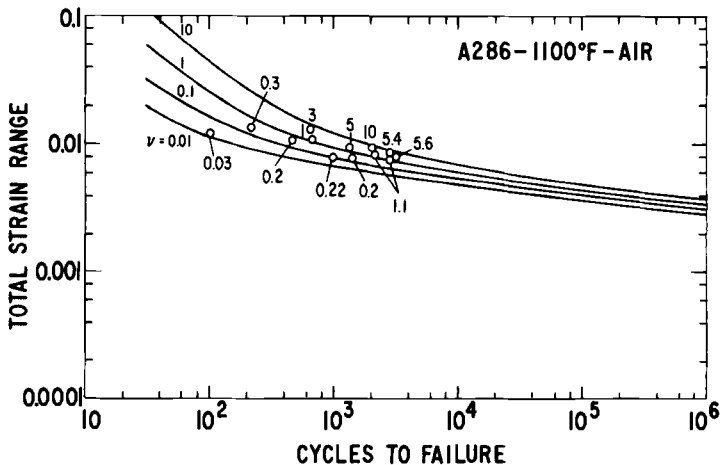
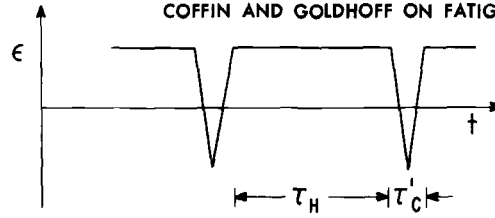
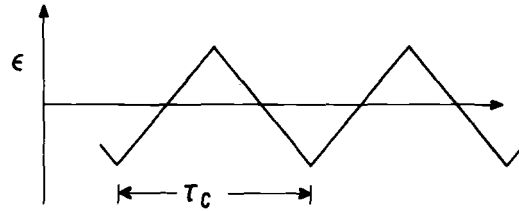


FIG. 33—Test results of total strain versus life as compared with analysis (solid lines) as predicted by Eq 28 for several frequencies. A-286 at 1100 F.



(b)



(a)

FIG. 34—Strain Program: (a) triangular wave form; (b) tensile hold.

tests. Thus if τ_c is the period of the triangular wave shape in Fig. 34a, then

$$\nu = \frac{1}{\tau_c} = \frac{1}{\tau'_c + \tau_H} \quad (23)$$

Berling and Conway recently published results of hold times on 304 stainless steel at 1200 F[113]. While their tests included equal hold times in tension and compression, compression hold times only, and tensile hold times, only the tensile hold time results will be considered here. Figure 35

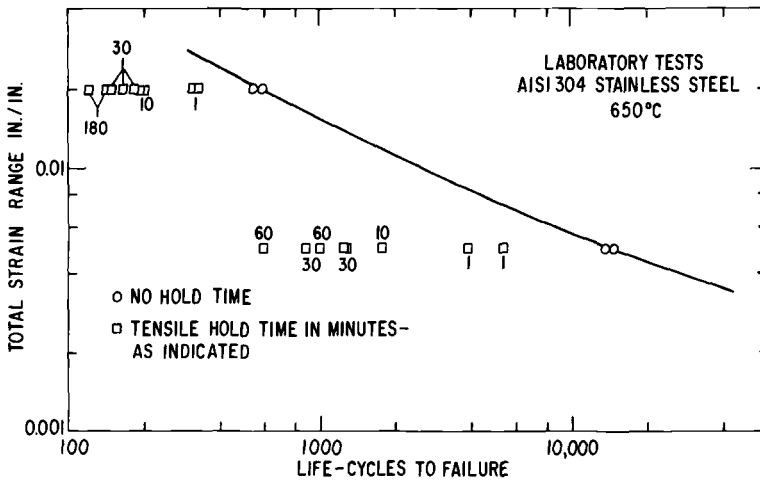


FIG. 35—Effect of hold time on life AISI 304 stainless steel at 650 C. After Berling and Conway[113].

shows their results for different tensile hold times compared to continuous cycling results at a strain rate of $4 \times 10^{-3} \text{ s}^{-1}$. A large decrease in life is noted, particularly for the longer hold times. To compare these results with those predicted by Eq 22, Eq 23 is assumed and the appropriate constants for the use of Eq 22 determined. Values of β , C_2 and k for this material were reported earlier[108], as obtained from the work of Berling and Slot[112]. Sufficient data is also given in that report to evaluate A , n and k_1 . These values are listed in Table 1. With these constants and the assumptions of Eq 23 the predicted life given in Eq 22 is compared to the test data of Berling and Conway. This comparison is shown in Fig. 36, both for hold time and no hold time test results. With the exception of one point, the error in predicted hold times results are within 30 percent of the actual lives. Other hold time comparisons are given in Ref 109.

Role of Environment in Time-Dependency

Very little work has been done on the fatigue behavior of metals in vacuum at elevated temperature. White[114] found that the fatigue life of a 0.5 percent molybdenum steel, subjected to continuous cycling at 1cpm in fully reversed bending, was twice as great in vacuum as in air at room temperature. Further the substantial reduction in fatigue life in air obtained by raising the test temperature to 500 C could be accounted for by oxidation, since vacuum tests at 500 C restored the life to those com-

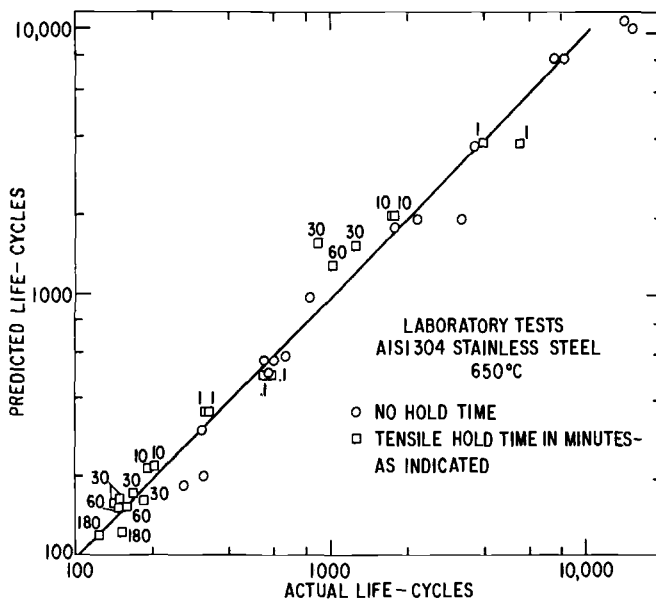


FIG. 36—Comparison of actual and predicted life with hold times—AISI 304 stainless steel—650 C.

parable to vacuum tests at room temperature. Additionally, it was found that the introduction of air into a continuous cycling test in vacuum at 500 C reduced the fatigue life by a larger factor than when a 30-min hold period per cycle was introduced into such a vacuum test. It was concluded that oxidation had a greater influence on fatigue life than creep. Figure 37 shows some of the findings of White's work.

The extensive work of Achter and coworkers as summarized in a recent publication[115] must also be cited. They investigated several materials, including nickel, steel and Inconel X, AISI 316 stainless at various temperatures up to 816 C in vacuums of 10^{-7} torr and various oxygen partial pressures in reversed bending. Increases in fatigue lives by as much as 100 times were found for the best vacuum for AISI 316 stainless steel,

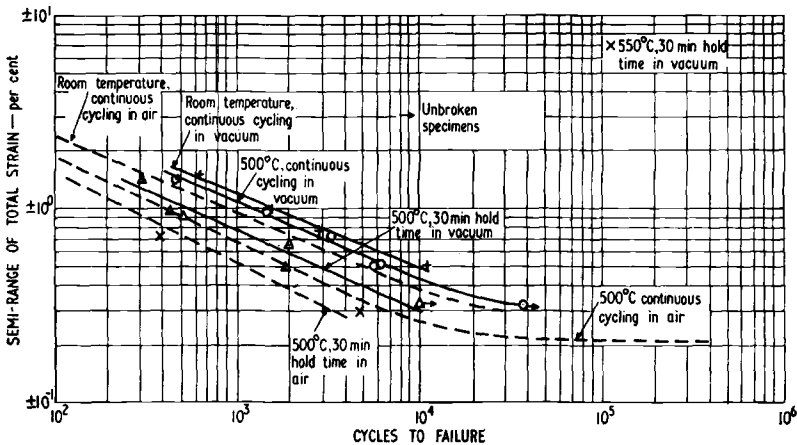


FIG. 37—Strain-cycling tests on $\frac{1}{2}$ percent Mo steel in vacuum and air, after White[114].

for example, dropping sharply for oxygen partial pressures greater than 10^{-4} torr. The effect of frequency at elevated temperature was not investigated.

The central question in relating the physical processes to the several phenomenological descriptions for time-dependency in high temperature fatigue is what process is controlling. From a metals engineering point of view, frequency, and hold time effects have been often identified with diffusional processes such as creep[116]. Two reasons can be cited for this: the stress relaxation effects that occur with extended hold times, and the transition in fatigue crack fracture mode from transgranular to intergranular[117].

On the other hand, evidence can be cited to support the position that the time-dependency of high temperature fatigue may be a result of the environment, and more specifically, oxygen. Not only is there the evidence

of several investigators, including White[114], Achter et al[115], and Nachtigall et al[118] showing the increase in fatigue life in vacuum, but also there is the tangible evidence that most fatigue cracks produced by slow cycling at elevated temperature are nearly always filled with oxide products. As indicated earlier, in an investigation of the damage and fracture mechanisms of cast Udimet 500 in high temperature fatigue, surface ridging of oxides at grain boundaries was found to be the source of crack initiation[91]. In A-286, a highly localized surface oxidation was identified as the initiation site for fatigue cracks under high temperature low cycle fatigue[119].

In order to clarify the dilemma of the importance of creep versus the environment in high temperature fatigue, an experimental program was undertaken to separate the two effects. This involves the determination of the frequency effect at elevated temperature both in air and vacuums of 10^{-8} torr. Thus, by eliminating the influence of the environment, the residual frequency effect must be due to creep processes or to phase instabilities. Some of the early findings of this program have been reported recently[120]. Shown in Fig. 38 are the comparative results of air and vacuum tests for A-286 at 1100 F and at several frequencies. Note the elimination of the frequency effect, and the substantial change in slope caused by testing in high vacuum. These experiments and similar results found on Udimet 500 at 1500 F more than confirm the observations of White. They show that for these two materials at the temperature of test, the environment is entirely responsible for the frequency dependence. If these findings continue to be confirmed by further experiments, some re-casting of our thinking with respect to life-prediction models may be necessary.

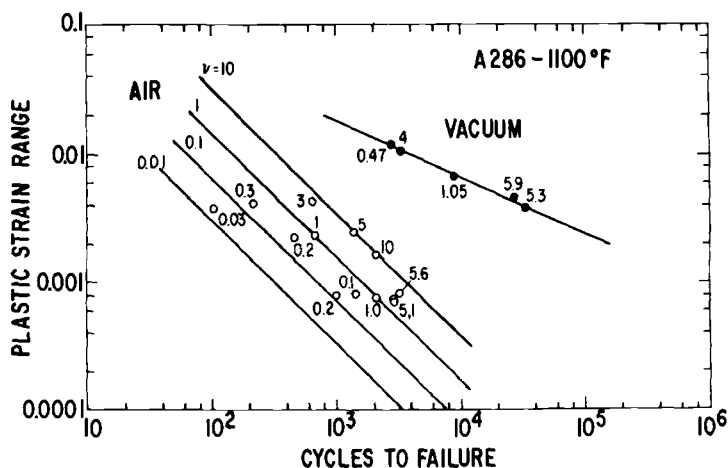


FIG. 38—Plastic strain range versus fatigue life for A-286 in air and vacuum. Numbers adjacent to test points indicate frequency in cpm.

Metallurgical Instabilities

Lazan[88] in discussing his findings on S-816 (Fig. 25) attributed the pronounced strengthening at high temperature to vibration induced solid state reactions. The work of Wells and Sullivan[121] and Coffin[122] give examples of pronounced cyclic strain hardening Udimet 700 and low carbon steels, respectively.

Following the position taken by Glen[123], strain-age hardening effects are largely responsible for the resistance to deformation for these alloys at elevated temperatures. Additions of carbon, nitrogen, manganese, molybdenum, and chromium produce strengthening depending on the degree to which they are in solid solution prior to deformation at elevated temperature. To increase creep resistance, these alloys are generally used in their normalized condition, whereby the cooling rates are such as to increase the degree of supersaturation of the alloy additions prior to use.

Creep resistance is attributed to precipitate formation and this is accentuated more by the combination of temperature and plastic strain than by temperature alone. It is clear that these effects are beneficial if the objective is to retard plastic flow at elevated temperature, while if the component is subjected to controlled cyclic plastic strain, these strain-induced reactions are "forced" to occur. Thus it is possible for cyclic plastic strain to accentuate premature failure. While a creep strain of 1 percent may be considered excessive in many cases, a much smaller but unaccounted for cyclic plastic strain occurring many hundreds of times can in fact limit the service performance of the component. For these materials there is an essential incompatibility in material selection between creep resistance and low-cycle fatigue resistance. Optimization of alloy composition and heat treatment to maximize creep resistance minimizes low-cycle fatigue resistance, while maximizing ductility for low-cycle fatigue capacity reduces the ability of the metal to withstand steady stresses at elevated temperature.

Notches

Relatively little work has been done on the elevated temperature low cycle fatigue resistance of notched bars. Of particular interest are the experiments of Krempl[124] on notched bars of three low strength structural steels, a 2.25 percent Cr — 1 Mo steel and AISI 304 stainless steel subjected to fully reversed loads. Since the notch root undergoes cyclic plastic strain, comparisons of the strain in this region with smooth bar data is suggested. Notch root strain measurements were compared with smooth bar fatigue results for equivalent lives. It was found that neither the axial strain range nor the effective strain range based on the octahedral shear strain or the maximum shear-strain correlated well.

An approach which follows that of Topper, Wetzel and Morrow[92] is

currently being studied at our laboratory. From Neuber's rule, it can be shown that

$$K_f \Delta S = (\Delta \sigma \Delta \epsilon E)^{1/2} \quad (24)$$

where ΔS is the nominal stress range applied to a notched member whose fatigue concentration factor is K_f , and $\Delta \sigma$ and $\Delta \epsilon$ are local stress and strain ranges in the notch root. By assuming conditions in the notch root to be equivalent to those in a smooth bar, the quantity $(\Delta \sigma \Delta \epsilon E)^{1/2}$ can be evaluated from smooth bar data for a particular life, and ΔS determined. In some cases the quantity K_f can be calculated from [125] as

$$K_f = 1 + \frac{K_t - 1}{1 + \frac{a}{r}} \quad (25)$$

where

r = root radius

a = material constant determined from long-life fatigue data for sharply notched specimens, and

K_t = theoretical stress concentration factor.

For high temperatures where time dependency must be considered it is attractive to combine our high temperature fatigue equations with the quantity $(\Delta \sigma \Delta \epsilon E)^{1/2}$. Rewriting Eq 24

$$K_f \Delta S = E[\Delta \epsilon_e^2 + \Delta \epsilon_e \cdot \Delta \epsilon_p]^{1/2} \quad (26)$$

Substituting Eqs 17 and 18 into 26, the quantity $K_f \Delta S$ can be expressed quantitatively as a function of life and frequency.

Guidelines for Further Research and Development

Rather than summarize the information presented in this paper, it might be useful to suggest areas where further work is needed to improve our capability to predict the behavior of materials subjected to high temperature creep and fatigue. These areas include:

Creep and Rupture

1. We need a more realistic understanding of the notch problem from a materials behavior standpoint, for example, when the failure conditions are limited by stress (relaxation phenomena) and when they are limited by strain accumulation. The pertinent strength and ductility prediction tests should be characterized so that we arrive at the proper compromise between strength and ductility in a given situation for commercial alloys.

2. A definition of a useful rupture criterion is required for multiaxial loading in creep based on initiation and growth of cracks. Predictive tests suited to such an investigation should be described.

3. A better definition is needed of the state of damage in a material using predictive tests so that (a) we know how much life is left after any history of exposure, (b) useful rules can be developed to quantify how life is used up, and (c) a threshold can be defined beyond which restorative heat treatments can no longer be usefully applied to exposed materials. (This implies a quantitative understanding of metallurgical (dynamic phase change) and mechanical (deformation and cracking) instabilities and their relation to one another so as to be able to predict life.)

4. There is a need for improvements in short time testing and related analytical techniques to insure more reliability in materials acceptance and predicting long time residual properties of materials exposed at elevated temperature.

High Temperature Fatigue

1. A better understanding of time dependency in high temperature fatigue is needed to provide guidance for design criteria.

2. A considerable amount of attention should be given to the development of improved high temperature fatigue design criteria including: (a) the development of design models based on crack propagation which includes plastic behavior and time dependency; (b) the development of criteria which better reflect the importance of environment in high temperature fatigue.

3. Improved analytical methods are needed for the treatment of complex geometries subject to cyclic loads, deformation, and temperatures.

4. Investigations of the effect of time dependent mean strain together with cyclic strain should be undertaken.

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An Assessment of Accelerated Testing

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ABSTRACT: The state of the art in accelerated life testing at the part level is assessed and found to be far less advanced than generally believed. Requirements and procedures to be followed in developing valid testing methods are presented. The use of a multidisciplinary approach is found to be essential.

KEY WORDS: predictions, accelerated tests, evaluation, components, tests, reliability, reviews, failure, stresses, statistical samples, statistical distributions, extrapolation

The Reliability Testing Problem

Today's aerospace systems are required to carry out increasingly complex and sophisticated missions. This usually results in the need for stringent reliability requirements which customers are increasingly wanting to see demonstrated prior to acceptance of the hardware.

The dilemma encountered here is that the greater a product's reliability the more difficult in terms of test time and test expense it is to demonstrate. These problems are not restricted to products of the aerospace industry however, but are constantly encountered by designers and manufacturers of all products regardless of their end use.

The designer has always been concerned with the integrity of the product manufactured to specifications which he has generated. The manufacturer requires knowledge of the quality and reliability of each production run as quickly as possible. For these reasons much effort has been exerted over the years across the spectrum of product types to develop test methods that furnish feedback to the designer and producer as quickly and as economically as possible to be used as objective evidence that the product of interest will perform its function satisfactorily for the specified period of time. The tests developed have been of many types. This paper

¹ Director, Reliability and Value Analysis, Bell Aerospace Co., Division of Textron, Buffalo, N. Y. 14240.

² Rome Air Development Center, Griffiss Air Force Base, N. Y.

deals with one of the major types of predictive testing—accelerated testing.

The research on which this paper is based was sponsored by Rome Air Development Center and the full results of the study are published as Refs 1 and 2.

The major objective of the study was to assess the state of the art of accelerated testing and to gather under one cover the methods and techniques that have proven useful or which appear to be potentially useful for reducing test times and expenses and yielding accurate repeatable results.

Definition of Accelerated Testing

An accelerated test method is one based on sound engineering and statistical assumptions, which utilizes a statistical model related to physical laws of failure to transform reliability information which has been generated in a relatively short time by an economical and accurate test method to quantitative repeatable estimates of a product's reliability characteristics had it been operated at rated stress levels.

This is a very complex definition, but is appropriate because it seeks to describe a very complex discipline. None of the aspects of the definition can be overlooked if one is to reap the benefits of reduced test times and accurate estimates of life expectancy. The literature on the subject of accelerated testing contains reports of much work that has been performed where one of the many complex facets of the above definition was ignored only to yield a whole set of study results that were less than satisfactory. Therefore, the next few paragraphs elaborate on the definition in order to stress the importance of strict adherence to the stated definition.

Sound engineering assumptions are required in order to select stresses and stress levels which will most likely result in the specification of a valid accelerated test method. Stresses that will not be experienced during actual use or which will change the physical state of a product's materials could but will not likely yield meaningful results.

Sound statistical assumptions will result in the generation of accurate, efficient, and above all, adequate quantities of test information to allow one to make reasonably accurate inferences regarding calculated results. One of the major statistical weaknesses encountered in the study of the literature of accelerated testing was the specification of inadequate sample sizes. Another major problem was the failure to utilize the benefit of the synergistic effects afforded by the application of several stresses simultaneously to induce failures in less time.

Other statistical requirements for valid accelerated test methods are that they should impart to the products tested, a cumulative failure distribution that results in failure more quickly over a given range of test time than other potential test methods. It should also produce a hazard

rate in the product tested that is higher at all points over a given range of interest than that of products operated at rated stress levels. The product tested at accelerated stresses should be described by the same general family of failure distribution functions as a product tested at rated stresses. While this latter point is not an absolute necessity, it must be recognized that the mathematical difficulties of transforming between results displaying different forms of failure distribution functions are not simple.

Once the correct statistical assumptions have been made, an accelerated test method requires a mathematical model or algorithm for converting failure times at accelerated stress levels to estimates of life expectancy if the product had been tested at rated stress levels. For the algorithm to yield accurate results, it must be based on accurate assumptions regarding the distribution of failure times.

Much of the work on accelerated testing has been based on the assumption that failure times are distributed exponentially at both accelerated and at rated stress levels. While this is an appealing assumption with regard to ease of calculations unless it is an accurate assumption, no meaningful repeatable predictive results can be expected.

The definition of an accelerated test refers to the desirability of being able to explain the reduced life expectancy at high stresses in terms of physical laws of failure. While it is clear that an empirically validated extrapolation algorithm can be used to estimate rated life expectancy based on accelerated test results without knowledge of the physical phenomenon occurring, it is still good assurance of accuracy and repeatability of the model if the results can also be explained in terms of failure physics.

A final aspect of the definition of an accelerated test method is that it should utilize accelerating stresses which are easily applied, controlled, and measured. Furthermore, equipment used in performing the test should be economical, reliable, accurate, and easy to adjust and maintain. If these principles are not adhered to, then test results may be clouded by errors in the generation of data.

Highlights of Rome Air Development Center (RADC) Research Study

Thus far this paper has dealt with the need for accelerated testing and the definition of the desirable attributes of such test methods. The remainder of the paper will discuss (1) the highlights of the RADC sponsored study program whose purpose it was to define and assess the state of the art of accelerated testing; (2) the opinions of the authors as to the course of research required to advance the state of the art; and (3) the empirical results of a test program conducted to develop and validate an extrapolation algorithm for a given part type.

As stated previously, much research to develop accelerated test methods has been carried on over a wide spectrum of products, stresses, extrapolation algorithms, and analysis methods. The degree of success in develop-

ing and implementing the resulting methods to yield accurate results also spans a wide spectrum. There was a need to assess the work that had been performed and to identify those accelerated test methods whose usefulness had been established and to unify the body of knowledge regarding valid test methods, models, and requirements. Therefore, RADC conceived and funded the research on which this paper is based.

The objective of the study was to search the literature of accelerated testing and extract from this mass of results those test methods and failure models which had been proven to be useful or those showing promise of further development. These results were then assembled under one cover as a reference text for the use of anyone concerned with the design, specification, or use of an accelerated test method. During the course of the research over 500 published articles on accelerated testing were reviewed and evaluated. The scope of the literature review is included as a bibliography in Refs 1 and 2.

The evaluation of the results of the literature review required a firm definition of the attributes an accelerated test must possess in order to be effective. The specific points used for the evaluation of the validity of each accelerated test method found in the literature were as follows:

1. there must be general acceptance of the statistical and engineering assumptions;
2. there must be a statistical model or algorithm for converting accelerated test results to estimates of life at rated stress levels;
3. there must be empirical proof that the statistical model yields accurate, repeatable results; and
4. there must be a physical model to explain the observed results in terms of laws of failure.

If only those accelerated test methods were considered valid which embodied all four of the above requirements, then there would be very little to report on the state of the art. Therefore, prominent examples of the traditional models are presented with some of the methods possessing as little as one of the above attributes.

Each method that has been included in the "Handbook of Accelerated Life Testing Methods"[2] is keyed to the electronic or mechanical part on which its use has been studied and reported in the literature. There are 25 such accelerated test methods and each is described in a standard format as shown below.

Format for Presentation of an Accelerated Test Method in the Handbook of Accelerated Life Testing[2].

Part Name and Description: Describe as fully as possible, that is: part number, manufacturer, nomenclature, rated loads, environmental ranges, material, size, application, etc.

Source: Publication(s) describing the method, author, date, source.

Purpose of Test: for example, compare parts, estimate MTBF, estimate reliability, estimate failure distribution, etc.

Degree of Validation: Describe the methods used and the success of these methods in validating the usefulness of the mathematical or physical models or both.

Description of Test Method: Include definition of failure, details of equipment used in testing, methods of gathering data, method of application of accelerating environmental or operating stresses, sample sizes, etc.

Summary of Results: Describe mathematical or physical model or both used, analysis methods, and general summary of results at both accelerated and normal stress levels, include failure mode observed. Include a numerical example, if necessary, for clarity.

Instructions for Use: Give a step by step description of how to use the method.

Limitations/Range of Applicability: Discuss any weaknesses of the method presented as to incomplete validation, questionable statistical or physical assumptions or use that can be made of the information presented. Name and describe the parts that can use this same method.

References: List references that apply directly to the method.

The purpose of this format was to aid the user of the Handbook in finding out as much information as he might require on how to use the method. The format tells him if a method is applicable for a given part type and outlines the general details required in terms of test equipment, stresses, stress levels, and sample sizes. He can also quickly determine the degree of validation of the method based on its previous use as reported in the literature. Results from other tests are included where available to give the potential user insight into expected results and approximate test durations. Step by step instructions for implementing the method are presented. They contain all the equations and models required to convert test results at accelerated stress levels to estimates of product life at rated stress levels. No references other than the Handbook should be required to utilize any of the methods presented. If a derivation or additional detail is desired by a potential user, he is furnished with the source from which each accelerated test method was abstracted. Limitations of use of the method and range of applicability accompany each method in the Handbook and inform the user of the weaknesses or risks of using methods not fully validated.

The specific methods described in the Handbook are those utilizing constant stress, progressively increasing stresses, step-stress, and the

Arrhenius model. Furthermore, the Handbook contains regression models, time transformation models, and a section on more powerful statistical methods (a body of analysis methods showing promise for advancing the state of the art of accelerated testing).

Major Types of Accelerated Tests

The following subsections summarize the general applicability, advantages and disadvantages of the major accelerated test methods described in the Handbook. The final subsection in this section presents the results of a validated time transformation model which has been found useful in reducing test times and expenses on relays of a certain design configuration.

Constant Stress Testing

The majority of accelerated test methods developed to date employ the features of stresses or overstresses or both applied at a constant rate. They can be applied at a single stress level or at many levels. The stress can be applied alone or in combination with other stresses.

If a single stress is applied at an accelerated stress level and also at the rated stress level, then the results of the longevities at the two levels form an estimate of the multiple of reduction of test time to be expected by application of an overstress. In this simple case no attention is paid to the distribution of failure times. The only assumption is that some constant acceleration factor can be used to convert accelerated stress life estimates at a given stress level to rated stress life estimates.

If more than one level of accelerated stress is employed, it becomes possible to plot a stress versus life curve (S-N Curve). This curve then represents the mathematical model displaying the manner in which life is varied by the application of different stress levels. This analysis method represents an added level of sophistication in the attempt to improve the accuracy of the estimation process. If the life expectancy at rated stress is too long to permit the performance of a test at this level, the S-N curve may be generated solely from overstress points and the curve extended or extrapolated to the rated stress using the slope or rate of change established from the several accelerated stress points on the S-N curve. It must hastily be pointed out that the extension of the model outside the range of empirical data is dangerous but it is sometimes the only course of action open. The magnitude of the risk is sometimes lessened if field data from previous experience has displayed results similar to those obtained from the extrapolation and also if these results indicate that the failure modes are similar at both accelerated and at rated levels.

The next level of complexity associated with the application of constant accelerating stresses comes when more than one stress is applied simultaneously. This may be necessary in order to generate an accurate simula-

tion of the manner in which a product is used during rated use. If it is subjected to a whole envelope of destructive stresses in normal service, then the application of some or all of them in combination at higher stress levels should reduce test time. More importantly, the interaction of two or more stresses may yield the dividend of an even greater rate of reduction in test time. Factorial experiments are used in the case of combined stress tests to identify the stresses or stress combinations that result in the most significant reduction in test time. Regression analysis can be used to develop the response surface representing the changing life characteristics brought about by the application of combined stresses.

The time transformation algorithms to convert empirical results at over-stress conditions to estimates of life expectancy at rated stresses can also be approached by utilizing certain assumptions regarding the distribution of failure times. For example if the distribution of failure times is exponential at both accelerated and at normal stress levels, then the time transformation between the two is a simple multiplier. In more complex cases where the distribution of failure times is considered and the exponential distribution does not hold, one can assume, prove empirically, or explain in terms of physical failure laws that some relationship exists between the distributions of failure times at rated and accelerated stresses. For example, it can be visualized that there could be a constant relationship between the accelerated and rated cumulative failure distributions, or a constant relationship between the accelerated and rated stress hazard rates or perhaps a ratio of the accelerated and rated hazard rates is a constant. If any of these relationships were found to exist, the result would be a mathematical model for converting accelerated test results to estimates of life at rated stresses.

The example appearing at the latter part of this section is a case demonstrating the situation where the distribution of failure times at both accelerated and rated stress level is Weibull and the validation of the algorithm is based on the assumption of a constant ratio of the hazard rates.

The application of stresses at a constant rate appears most frequently in the literature across the total spectrum of part types. The advantages of the use of this type of stress in accelerated tests lies in the fact that it is possible to construct a distribution of failure times and also analysis methods can be utilized to evaluate both the main and synergistic effects which operate to reduce test time. The application of the stresses at a constant rate also allows the construction of the S-N curve relating stress and part life.

The disadvantages of applying stress at a constant rate are that tests conducted at intermediate and low stress levels tend to be excessively long. Directly related to this is the fact that if the dispersion of failure times is large within a given stress level, it is possible to experience either very

short or very long test times. At high stress levels, failures might occur almost instantaneously thus making the estimation of parameters difficult. On the other hand, an outlier lasting a long time could greatly increase calendar test time unless some kind of censoring plan was adopted. Furthermore, it might be necessary to carry on a series of preliminary screening experiments to select the optimum stress level before embarking on the final test plan.

In summarizing the state of the art it appears that accelerated tests with the stresses applied at a constant rate and in combination with other stresses appear to offer the most promise for an efficient method of generating the most accurate results.

Step Stress Testing

Step stress testing is an accelerated test method in which one begins with a fairly large lot of specimens and tests for a fixed time interval beginning at a fairly low stress level. At the end of the time period, the stress is increased and the good parts remaining from the previous step are subjected to the increased stress for the same time period. Again the process is repeated of taking the remaining part of a test lot, increasing the stress and testing for a fixed time period until some fraction of the lot of parts has failed.

The method has been employed mainly on electronic parts with the most work reported in the literature being recorded on semiconductors, capacitors, and resistors.

The step stress method is based on some very important assumptions as follows:

1. The S-N curve can be transformed into a straight line by some transformation representing a physical model. Usually the transformation is logarithmic on the time scale and the reciprocal of the absolute temperature on the stress scale.

2. The failure times at any point on the S-N curve must be normally distributed with equal variances.

3. The transformed S-N curve represents the effects on life of a single failure mechanism or at least of a dominant failure mechanism.

4. The probability of failure at any point in the stress-time domain is independent of how a part arrives at that point.

The advantages of the method are that it solves the problem of excessively early and excessively late failures in accelerated testing. It is effective in pinpointing the stress level at which failure mode changes occur. It also is effective in comparing the life characteristics of two different part types or two different design features in a qualitative way.

Its disadvantages are that it has not been proven valid for quantifying reliability, and it requires a good deal of monitoring and care to perform the tests at exact stress levels for exact time intervals. This monitoring

and administrative detail are necessary if the test plan is to yield useful results.

Progressive Stress Tests

This form of accelerated test has been used almost exclusively on capacitors. It appears to be similar to the step stress test, but instead the stress is applied at a low level and is increased at a given rate until the desired fraction of the test lot has failed. It is used in conjunction with the inverse power rule and is a cumulative damage model unlike the step stress model which assumes no damage at lower stress levels.

There is not much evidence of validation of the model, but it has been widely used. Most authors in the literature used the assumption of exponential failure times in both constant stress and progressive stress tests to construct the damage model. It has been proven, however, that if failure times at constant stress are exponential, then failure times at progressively applied stresses must be distributed according to the Weibull with a shape parameter $n + 1$ where n is the power exponent of the inverse power rule.

The advantage in the use of this type of accelerated test method is that it eliminates the typical problem of early and late failures. Its disadvantages are that it has not as yet been validated using the inverse power rule and it also requires stringent controls to assure that the rate of application of the stress is accurate.

The Arrhenius Model

The Arrhenius Model is based on a physical relationship that has been found to describe the reaction rate of chemical processes. Since the application of accelerated stresses may result in degradation through higher temperatures, the use of the Arrhenius Model in explaining the results obtained seems logical.

Its use in accelerated testing consists of searching for some performance parameters whose degradation rate is linear with time. If none can be found, then various transformations such as logarithmic, square root, or others are employed to produce a straight line. When such a parameter is found, failure data for this parameter is collected from several different stress levels. The slopes of the linear plots at the various stress levels are then plotted against the reciprocal of the absolute temperature. If these Arrhenius plots are linear, then true acceleration is said to exist.

This model has been used to develop accelerated tests on many types of electronic parts. In most cases it appears that an insufficient number of stress levels have been utilized to yield validation of the model. A convenient way of checking for whether or not this model fits is to test the distribution of failure times. If it is exponential at accelerated stresses, it must also be exponential at rated stress. If the distribution is Weibull at

accelerated stresses, it must also be Weibull with the same shape parameter at rated stress.

History and Example of a Time Transformation Accelerated Test Method

An investigation was conducted over the time span of 1965 to 1969 of accelerated life testing of relays. More specifically, a crystal can relay of the type MIL-R-5757/10 RY4NA3B3 LO1 was selected due to its considerable usage and large number of qualified products' list (QPL) sources. The relay is double pole, double throw, contact rating of 3 A at 28 V(dc), 1 Hz actuation, temperature range -65 to $+125$ C. The coil is rated at $675\ \Omega$ and 26.5 V.

The complete details of the investigation are presented in Refs 3, 4, and 5. The significant findings of the investigation are as follows:

1. the lives of the relay follow the Weibull distribution at normal and overstress operating conditions with failure defined as first miss;
2. the optimum stress conditions for accelerating failures are 6 A contact current, 30 Hz actuation rate, and 25 C temperature; and
3. the accelerated test results can best be converted to estimates life at rated conditions by an algorithm which assumes a constant ratio of the Weibull hazard rates. The algorithm takes the following form to provide the estimates of the parameters of the rated stress lifetime two parameter Weibull cumulative distribution:

$$\alpha_N^* = \left(\frac{\alpha_N \beta_A}{\alpha_A \beta_N} \right) \left(\frac{\alpha_A^* \beta_N^*}{\beta_A^*} \right)$$

$$\beta_N^* = \beta_A^* + (\beta_N - \beta_A)$$

where:

α_N^* = estimate of Weibull scale parameter for parts if they had been tested at manufacturer's rated operating and environmental stresses;

β_N^* = estimate of Weibull shape parameter for parts if they had been tested at manufacturer's rated operating and environmental stresses;

α_A^* = estimate of Weibull scale parameter obtained from a current test run at accelerated stresses;

β_A^* = estimate of Weibull shape parameter obtained from a current test run at accelerated stresses;

α_N = estimate of Weibull scale parameters from a previous test run of parts operated at manufacturer's rated conditions;

β_N = estimate of Weibull shape parameter from a previous test run of switches operated at manufacturer's rated conditions;

α_A = estimate of Weibull scale parameter from a previous test run of switches operated at accelerated stress conditions; and

β_A = estimate of Weibull shape parameter from a previous test run of switches operated at accelerated stress conditions.

The following inputs from the investigation illustrate the relative accuracy of the algorithm:

$$\begin{array}{lcl}
 \alpha_A^* = 3.16 \text{ E } 6 & \left. \vphantom{\alpha_A^*} \right\} & \text{Sample Size 10} \\
 \beta_A^* = 1.13 & & \\
 \alpha_N = 5.7 \text{ E } 22 & \left. \vphantom{\alpha_N} \right\} & \text{Sample size 50} \\
 \beta_N = 3.79 & & \\
 \alpha_A = 1.53 \text{ E } 6 & \left. \vphantom{\alpha_A} \right\} & \\
 \beta_A = 1.1 & & \\
 \beta_N^* = 1.13 + (3.79 - 1.1) = 3.82 & & \\
 \alpha_N^* = \left(\frac{5.7 \text{ E } 22 \times 1.1}{1.53 \text{ E } 6 \times 3.79} \right) \left(\frac{3.16 \text{ E } 6 \times 3.82}{1.13} \right) = 1.16 \text{ E } 23. & &
 \end{array}$$

A random sample of 10 from the same lot that the accelerated lot was selected from was run at the manufacturer's rated conditions of three amps contact current, 1 Hz actuation rate, and 25 C ambient temperature so that a comparison of the Weibull parameters could be made. The 95-percent confidence interval was also constructed about these rated stress test parameter estimates. The comparison is given below:

<i>Algorithm</i>	<i>Rated Stress Test</i>
α 1.16 E 23	$\left\{ \begin{array}{l} (.975) \text{ } 6.5 \text{ E } 33 \\ 1.73 \text{ E } 17 \\ (.025) \text{ } 8 \text{ E } 9 \end{array} \right.$
β 3.82	$\left\{ \begin{array}{l} (.975) \text{ } 5.40 \\ 2.73 \\ (.025) \text{ } 1.62 \end{array} \right.$

It can be seen from the results that the algorithm provides quite usable results.

In summary it appears that the time transformation model has been validated empirically and can be utilized successfully on relays of the type which were used in this series of tests.

Strictly speaking however it must be said that the requirements of the definition of accelerated tests presented earlier in the paper have not all been fulfilled by this method. For example, the validation has been accomplished based on empirical results. No physical failure laws have yet been found to explain the empirical results observed in the several test programs. Furthermore, it must be noted that to use the algorithm it is necessary to have one set of test results from tests at rated conditions. This limitation is severe if the algorithm is to be applied to a new product

where no application data is available or if a product has an exceedingly long life expectancy (say 5 to 10 years). On the other hand for parts with relatively short lives, it is possible to generate a set of normal stress data and in these situations, the test method appears useful.

It requires further study to determine if it can be applied to other relay types and to other part types whose failure times are distributed according to the Weibull.

The State of the Art

The paper up to this point has discussed the need for accelerated testing, has defined the factors necessary to constitute a valid accelerated test, and has described the general types of test programs attempted in the past with varying degrees of success regarding validation. In summarizing the state of the art, it must be said that there is not evidence in the literature that any of the methods attempted have been validated in conjunction with the definition of accelerated testing which was developed in the research study and which was presented earlier in this paper. This is not to say that no useful effort has evolved from the monies expended, but it does indicate that the discipline of accelerated testing has fallen short of the expectations of most people for several clearly identifiable reasons. The authors' opinions of these reasons are as follows:

1. The principles involved in designing and conducting a valid accelerated test method are so complex that they cross the lines of many disciplines such as physics, statistics, electrical/mechanical engineering, and reliability engineering to name but a few. The literature of accelerated testing displays that most work has been conducted by members of usually only one of these disciplines. Therefore, the test methods which have been designed primarily by members of only one of the aforementioned disciplines have not taken into consideration important requirements dictated by the other disciplines. The result has been test methods yielding results that are less than perfect.

2. Generally the statistical approach in the development of accelerated test methods was weak. This took the form of sample sizes that were too small to yield statistical significance, insufficient numbers of observations to fit regression equations, and the frequent use of the assumption of failure times distributed exponentially without verifying the assumption or where the assumption did not apply.

3. There was little evidence that much effort was made to validate the methods developed by attempting to obtain repeatable results.

4. The difficulty of carrying out complex mathematical calculations required by the use of certain models or failure distribution assumptions has led to the use of simplifying assumptions resulting in inaccurate results.

The Future of Accelerated Testing

The methods developed and tried thus far have not yielded completely validated results. Therefore, if the state of the art is to improve and produce more useful results, the following steps must be taken:

1. Greater attention must be paid to the design of the test method. This statement includes not only the statistical aspects of proper sample sizes, failure distribution assumptions and the like, but calls for the utilization of the interdisciplinary approach.

2. New statistical tools and analysis methods should be investigated. For example, much work is being done in developing analysis methods using the Weibull distribution, analysis methods using censored data, Bayesian statistics, nonparametric statistical methods and competing risks failure models. This whole area of "More Powerful Statistical Methods" is reviewed in Refs 1 and 2. It is evident that advances in theory are being made which when coupled with the added power of computers in complex problem solution should result in significant improvements in the accuracy of predictive testing methods of the near future.

3. More work must be done in searching for physical explanations for the failure phenomena which are observed in accelerated tests. Furthermore, it is imperative that more effort be placed on the validation of the models which are developed and used.

4. Most of the research done thus far in the field of accelerated testing has been performed at the part level. Future attention must be directed toward the development of methods that will yield predictive results on equipments and systems.

5. Additional effort must be devoted to time transformation models to translate the methods validated thus far (an example of which is presented in this paper) to other generic part types.

6. Advantage must be taken of certain statistical facts which can be used to determine if a given failure model can possibly be utilized. As an example, it is a fact that in the use of the Arrhenius model if the distribution of failure times at accelerated stress levels is exponential, then the distribution of failure times at rated stress levels must also be exponential or else the model cannot be considered valid. Certain other combination of distributions of failure times such as Weibull at accelerated stresses and exponential at rated stresses are simply not a possible situation. Therefore, this principle can be used to determine the validity of the assumption that this model is applicable.

In conclusion it must be said that the technology of accelerated testing still appears to hold much promise for development as a predictive tool. Progress is being made, more realistic models are being applied and more powerful analysis methods are being developed and employed. Although

the ratio of success to effort enjoyed in the past has not resulted in outstanding results, the signs of development are evident. Even though no major breakthrough has been noted in the time period from the publication of Refs 1 and 2, one has only to review the current literature to realize that more powerful statistical methods and the search for physical failure models is progressing and will eventually result in substantial improvements in the utility of accelerated testing as a predictive tool.

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The Evaluation of Creep Damage in a Cr-Mo-V Steel

REFERENCE: Goldhoff, R. M. and Woodford, D. A., "The Evaluation of Creep Damage in a Cr-Mo-V Steel," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 89-106.

ABSTRACT: An attempt has been made to determine and quantify the accumulation of creep damage in Cr-Mo-V steel. Damage is defined in terms of post exposure rupture data and hardness measurements which are shown to be interrelated. Correlation between the post exposure rupture life and the initial exposure parameters are achieved by simple parametric techniques which then form the basis for graphical representation of damage accumulation as a function of exposure time, stress, and temperature.

The way in which these curves may be used to predict failure under nonsteady conditions of stress and temperature is described. In general the linear life fraction cumulative damage rule is unrealistic for this material, although under certain restricted conditions it may give a reasonable prediction of life. Apart from the measured nonlinearity with time of damage accumulation, the present results confirm that the common assumption of a unique failure criterion independent of the test conditions cannot be justified.

KEY WORDS: predictions, plastic deformation, creep properties, precipitation hardening, softening, stresses, low alloy steels, deformation, creep rupture strength, damage, crack propagation, cavitation corrosion

All machinery operating at elevated temperature experiences cyclic conditions. This may range from simple stop-start cycles with the intervening operation at relatively constant conditions to a random variation of the important variables constituting very complex cyclic behavior. The engineer would like a simple analytical means for assessing cyclic damage to test the reliability of a design for which cyclic parameters can be estimated and for which certain types of materials test data are available. Clearly, there is a need to examine, experimentally and theoretically, the possibility of predicting behavior under nonsteady temperatures and stresses.

To tackle these problems, the concept of cumulative damage has been

¹ Manager, Applied Research and research metallurgist, respectively, Materials and Processes Laboratory, General Electric Co., Schenectady, N. Y. 12305.

introduced. Plastic deformation under any test condition is supposed to change the structural and mechanical state of the material; the response to a change in test conditions then depends on the prevailing state. In the simplest approach, damage is described in terms of the time under load so that, relative to the damage at failure in a creep rupture test, the fractional damage after time t is t/t_R where t_R is the corresponding rupture time. Under conditions of incremental stress changes, assuming that each fractional expenditure of life or accumulated damage, is independent of all others, failure is predicted when the sum of the fractions of life equals unity[1]

$$\sum_{i=1}^n \frac{\Delta t_i}{t_{Ri}} = 1 \quad (1)$$

where

Δt_i = time spent at stress σ ,

t_{Ri} = rupture time at constant stress σ , and

n = number of stress steps.

The above description of creep damage is really no more than a "rule of thumb" and, in general, will not be applicable to other than an extremely restricted range of test conditions. It is believed that a realistic description of flow and fracture under nonsteady conditions must be based on an understanding of the effect of deformation on the components of damage.

Apart from surface changes such as decarburization, oxidation, and corrosion, the major property changes resulting from creep exposure may be related to three types:

1. the creep deformation;
2. thermal and strain induced precipitation phenomena;
3. initiation, growth, and propagation of grain boundary cavities and cracks.

By the first type of damage we refer to changes in dislocation density and arrangement commonly examined in simple metals and alloys. In complex alloys there may also be substantial changes in precipitate size, type and distribution. The effect of the second type of processes on embrittlement and consequent structural damage has recently been reviewed[2]. Except in the case of those high purity metals which do not show grain boundary cracking (Type 3), all three damaging processes will be expected to occur to varying degrees.

To predict behavior after prior exposure we need to define the current state of the material either directly in terms of a quantitative microscopic description of the damage components, or indirectly in terms of a property measurement sensitive to creep damage. We have chosen the latter approach in view of the complexity of microstructural changes experienced in a commercial alloy and the consequent inability to quantify

them. Unfortunately, the total effects of creep damage cannot, in general, be described uniquely in terms of the subsequent response to any single mechanical property. For example, the low temperature impact properties might be especially sensitive to the presence of creep cracks (Type 3), whereas the yield strength and work hardening characteristics would be anticipated to depend more on the prevailing dislocation and precipitate structure.

To circumvent these difficulties we have used post exposure rupture life as a measure of creep damage. The rationale is that a high temperature creep-to-rupture test is most likely to describe the damage state relevant to high temperature creep deformation. In other words, it will be sensitive to the combined effects of the three damaging components described. There were two major objectives in this work:

1. to establish a correlation between a measure of damage and the initial exposure conditions; and
2. to develop a method to predict failure under nonsteady temperatures and stresses based on damage accumulation.

Experimental Details

The composition, heat treatment and mechanical properties of the Cr-Mo-V steel are described in Table 1.

Cylindrical creep specimens with a 7-in. gage length and 0.357-in. diameter were tested at four temperatures and various stresses. Temperature control was to ± 2 F, load control within 0.5 percent of required load, and eccentricity on loading 8 percent or better. Half of the tests were stopped prior to failure. A single rupture specimen, 1.25-in. gage length and 0.160-in. diameter, was machined from the uniform strain section of each exposed bar and subjected to a constant load test at 1000 F and 35 000 psi.

To complement this rupture test, which limits the damage assessment

TABLE 1—Composition, heat treatment, and mechanical properties of test steel.

Chemical Composition								
C	Mn	P	S	Si	Ni	Cr	V	Mo
0.30	0.92	0.030	0.031	0.34	0.08	1.16	0.23	1.22
Heat Treatment								
Austenitize:								
1850 F, 27 h, air cool								
1750 F, 27 h, air cool								
Temper and Stress Relief:								
1250 F, 36 h, furnace cool								
1150 F, 36 h, furnace cool								
Mechanical Properties at R. T.:								
Tensile Strength, psi		Yield Strength, psi		Elongation, %		Reduction of Area, %		
112 000		86 250		18.5		47.2		

TABLE 2—Original and post exposure data on Cr-Mo-V steel, item 5031B.

Bar Number	Original Exposure				Post Exposure Creep ^a					Hot Hardness				
	Temp., deg. F	σ , 1000 psi	t_e , h	ϵ_e , %	RA_e , %	t_p , h	ϵ_p , %	RA_p , %	$\% / h$ MCRp	Int. P25	Int. P30	Slope	2000 s 1100 F	2000 s 2000 F
2A (orig.) ^b	...	...	...	...	...	973	19.4	82	.0080	149.9	87.5	.0125	89	122
2C	900	45	4 306	18.7	65.5	286	21.7	73	.018	135.6	78.4	.0115	79.5	106
2O	900	35	60 376	8.7	44.0	138	16.4	50	.028	118.2	76.9	.0083	80	101
2W ^c	900	25	48 063	0.33	...	313	20.7	63	.013	129.7	80.0	.0100	76	104
2T ^c	900	15	31 817	0.09	...	540	18.0	76	.008	141.9	79.6	.0124	84	108
2AA ^c	1000	37	24	.65	...	895	21.3	75	.0038	...	...	...	...	...
2BB ^c	1000	37	118	1.4	...	546	30.1	92	.0068	...	...	...	...	...
2CC ^c	1000	37	216	2.1	...	574	23.4	77	.0046	139.4	80.0	.0119	82.4	104
2E	1000	30	4 663	9.8	71.0	164	19.5	56	.031	134.8	79.6	.0111	72.0	108
2R	1000	25	29 713	6.8	34.4	54	16.3	56	.084	116.3	74.0	.0085	75.0	96
2DD ^c	1000	23	1 510	0.37	...	443	26.5	77	.011	133.5	76.0	.0115	66.5	98
2EE ^c	1000	23	2 331	0.41	...	402	21.9	78	.011	138.3	80.4	.0116	66.5	107
2Q ^c	1000	15	59 171	0.60	...	122	21.2	74	.043	124.7	73.4	.0103	73.0	99
2V ^c	1000	10	37 042	0.17	...	179	17.6	77	.026	129.8	72.3	.0115	73.0	101
2F	1050	25	3 205	9.8	72.5	74	26.3	81	.072	116.5	70.7	.0092	64.0	93
2G	1050	20	17 860	7.6	54.4	50	22.8	54	.096	116.5	65.5	.0107	63.0	84
2V	1050	15	63 870	7.8	35.2	11	21.7	45	.70	110.7	62.0	.0098	61.0	89
2X ^c	1050	10	53 521	0.8	...	46	30.0	77	.086	...	...	...	...	...
2I	1100	20	2 825	10.1	74.0	31	28.1	74	.17	112.3	67.9	.0089	67.0	95
2J	1100	15	11 055	12.5	52.5	12.3	30.9	89	.78	109.4	65.7	.0087	65.0	87
2H	1100	10	32 785	9.2	70.4	4.8	31.7	69	1.96	105.5	65.5	.0080	64.0	88.5

^a All tests at 1000 F and 35 000 psi.^b Average of three tests.^c Did not fracture.

to one test condition, a series of hardness measurements were made on a section parted from the exposed bars adjacent to the rupture test blank. Temperatures of 900, 1000, 1100 and 1200 F were used with indentation times of 20, 120, 600 and 2000 s. Rupture and hot hardness measurements were also made on unexposed material for base data.

A complete summary of the test specimens used in this analysis, their initial exposure conditions, and all the post exposure damage measurements are shown in Table 2. These represent only a portion of the total specimens in the complete data set which is fully documented in Ref 11.

Correlation Of Results

Creep-Rupture Tests

The first objective is to relate the post-exposure test results to the initial exposure data. From Table 2 it is clear that the time to rupture in the post-exposure test, t_p , is more sensitive than either elongation or reduction in area. The effect of initial exposure on this parameter was therefore examined.

Figure 1 shows the correlation between t_p and t_o (Subscript o henceforth denotes initial exposure data). Although a general trend of temperature and time is apparent, the overall correlation is poor. On the basis that both strain and time appear to be important in producing

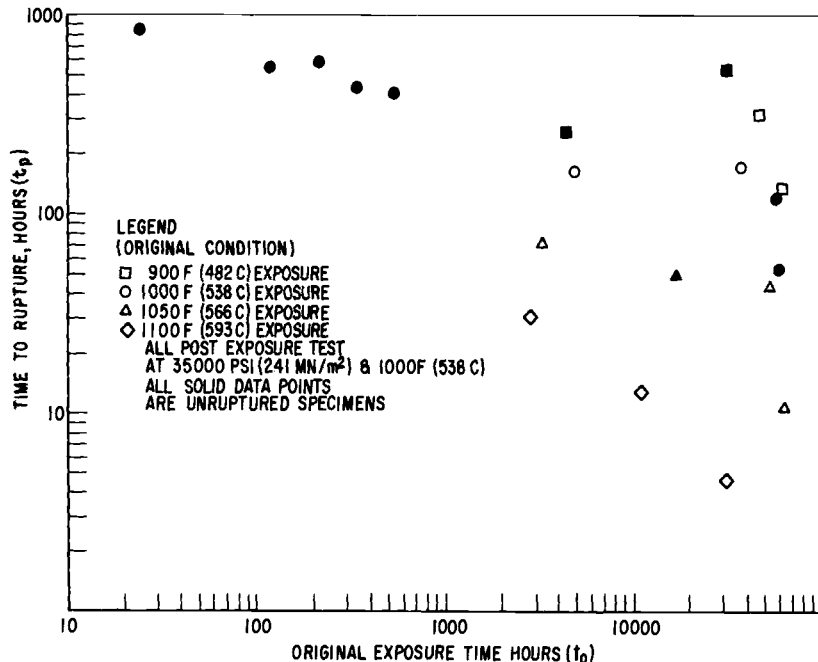


FIG. 1—Correlation between post exposure rupture times (t_p) and original exposure time (t_o).

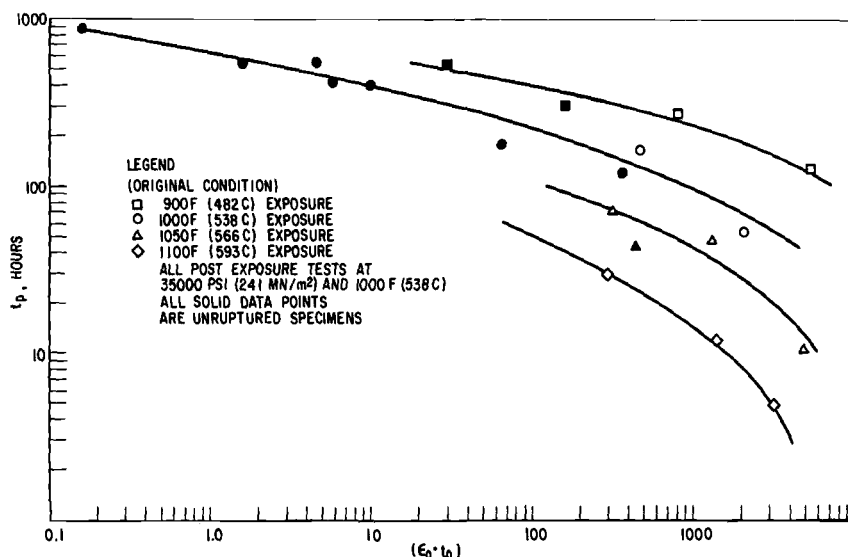


FIG. 2—Correlation between post exposure rupture time (t_p) and product of original exposure strain and time ($\epsilon_o \cdot t_o$)

damage[3-5], and that the total volume of cavities in a pure metal (Type 3 damage) is directly proportional to the product of the strain and time[3,4], Fig. 2 was constructed. Here ϵ_o is either creep strain in unruptured bars or rupture elongation in fractured specimens² and a good correlation should not be taken to imply that the principal damaging process is one of cavitation. Indeed as will be shown, grain boundary cracking appears to play a minor role in this alloy. However, it does demonstrate that both strain and time are important and represents an empirical correlation with the product of strain and time. Having established this product as a basis for damage correlation it was found that a plot similar to Fig. 2 could be drawn using post-exposure minimum creep rate but not ductility, that is, ϵ_p or RA_p . Combinations involving the ductility, for example, ϵ_p/t_p or $\epsilon_p \times t_p$, were found to yield reasonable correlations but not as good as that shown in Fig. 2.

The data of Fig. 2 were analyzed by simple graphical techniques[6] and yielded a reasonably good parameter fit of the universalized Manson-Succop[7] type shown in Fig. 3. The ordinate in this figure is shown as "damage ratio" which is the simple ratio t_p/t_c where t_c is the mean rupture time of the original alloy for three tests at the postexposure test conditions of 35 000 psi and 1000 F, that is, 973 h. The master curve can be best represented by the polynomial:

² The use of rupture elongation rather than the uniform strain of the unnecked portion of a failed bar introduces only a small error because: (a) the creep bars had a 7-in. gage length so the effect of necking on the total strain is small, and (b) for this alloy necking does not begin until the very last stages of life, that is, well into the tertiary creep stage; hence, the uniform strain and rupture strain may be quite close in value.

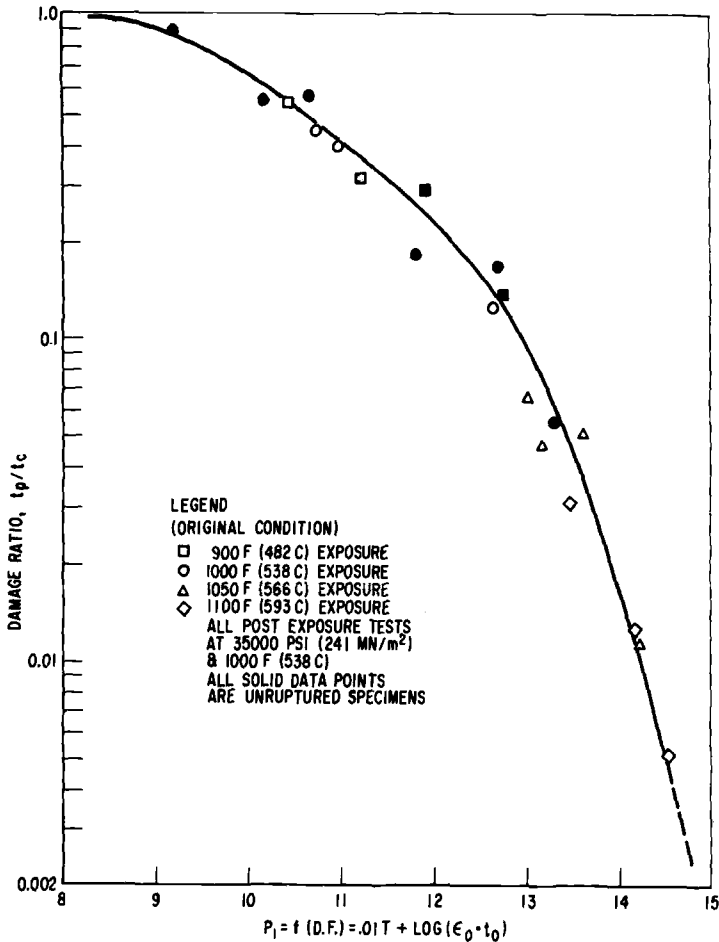


FIG. 3—Master curve for the parameter representation of creep damage.

$$t_p/t_c = -7.49 + 1.54P_1 - .081P_1^2 \quad (2)$$

(Index of Det. = .972)

where $P_1 = .01T + \log(\epsilon_0 \times t_0)$ for T in F.

Figure 3 provides an estimate of t_p for any combination of ϵ_0 , t_0 and T within the range covered. It should be noted that stress does not appear directly but is defined implicitly for a given combination of the three exposure parameters. To relate the initial stress we may make use of the regular temperature dependence of a plot of σ_0 versus ϵ_0/t_0 . Using the same techniques as before, the parametric plot of Fig. 4 was constructed which is linear and is easily described:

$$\sigma_0 = 1.78 \times 10^5 + 6540 P_2 \quad (3)$$

where $P_2 = f(\sigma_0) = \log \frac{\epsilon_0}{t_0} - .02T$

Obviously we now have a functional dependence of both t_p/t_o (Eq 2) and σ_o (Eq 3) on the original exposure parameters ϵ_o , t_o and T . By solving for any of these parameters in Eq 3, and substituting in Eq 2, we can incorporate a stress dependence for the damage accumulation. Such an expression is cumbersome to use, and it is simpler to use the master curves of Figs. 3 and 4 directly to provide a useful description of damage accumulation. This is shown later in the paper.

Hot Hardness Tests

One of the limitations on the postexposure rupture test is that all specimens were tested at only one condition; namely, 35 000 psi and 1000 F. A question arises as to the likelihood of correlations similar to those shown for other post exposure conditions. Since exposed test ma-

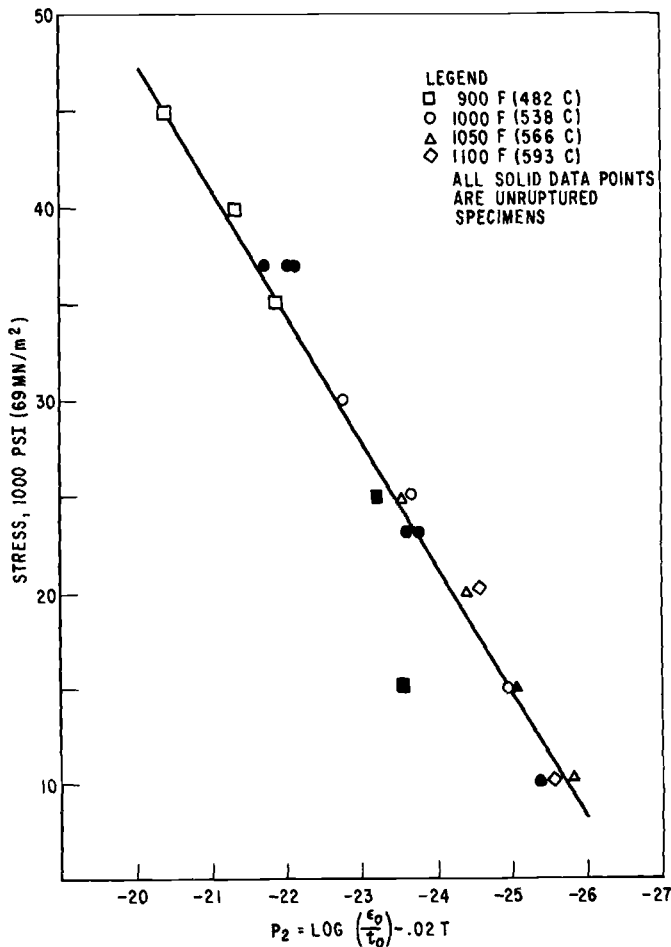


FIG. 4—Master curve for the parameter representation of average creep rate ϵ_o/t_o as a function of stress, σ_o .

material was not available in sufficient quantity for additional creep testing, the time dependent hot hardness survey was used to supplement the program. Plots of hardness versus $\log t$, where t is the indentation time in seconds, were analyzed to produce a master curve again based on the Manson-Succop parametric form: $P = AT + \log t$, where T is the temperature in deg F. Approximately one-third of the data was treated in this way and an average value of 0.024 was found for the constant A . Data for each specimen were then plotted as a function of this parameter as shown for specimens 2A and 2J in Fig. 5. Similar plots were obtained for other specimens and the linear relation allows for direct comparison between exposed materials using either hardness at fixed parameter numbers or the slope of the master curves. Listed in Table 2 are the least squares

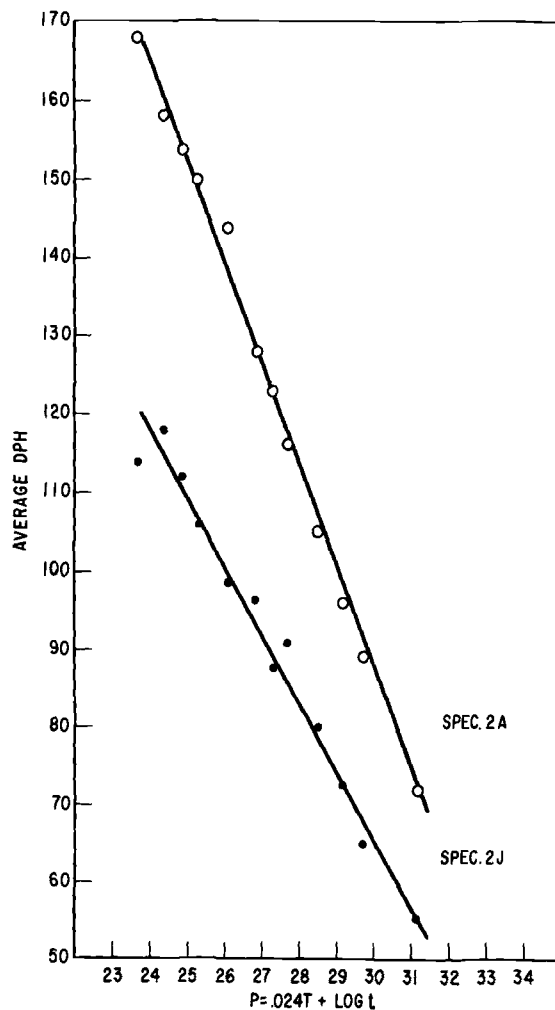


FIG. 5—Master curves for the parametric representation of hot-hardness data.

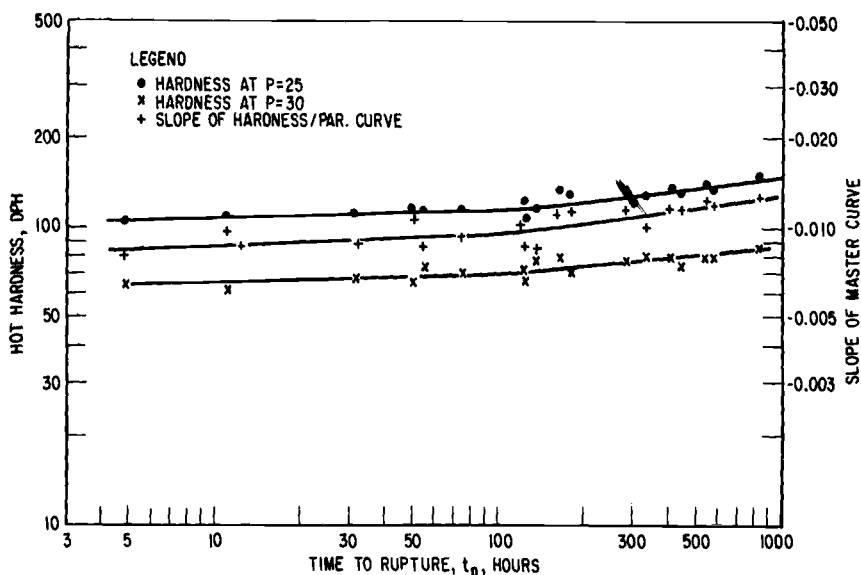


FIG. 6—Correlation of hot hardness characteristics and post exposure rupture data.

fitted values of hardness intercepts at parameter values of 25 and 30 and the slope of each master curve for each specimen tested.

In Fig. 6 the correlation between the tabulated hot hardness characteristics and the corresponding post exposure rupture times are shown. It is evident that reasonable correlations do exist, particularly for the various intercept values. However, the hot hardness data are far less discriminatory. The diverging lines in Fig. 5 indicate that a greater differentiation between the damaged and undamaged condition would be obtained by using hardness data at lower temperatures and the almost constant ratio of hardness also apparent in this figure would indicate that a hardness value obtained at low temperature would be indicative of the high temperature strength. Figure 7 was therefore constructed which shows a reasonably good correlation between t_p and room temperature hardness.

These data indicate that hardness testing is an effective way to assess elevated temperature damage in this alloy. They also imply that softening with exposure is the major damaging process.

Formulating Damage Criteria

The chief interest in this work is to devise a method of predicting failure not only under static loading conditions, but also under conditions of varying stress and temperature. Ultimately, a method that applies to fatigue loading at elevated temperature would be desirable. We initially consider the simplest concept commonly in use which is shown schemati-

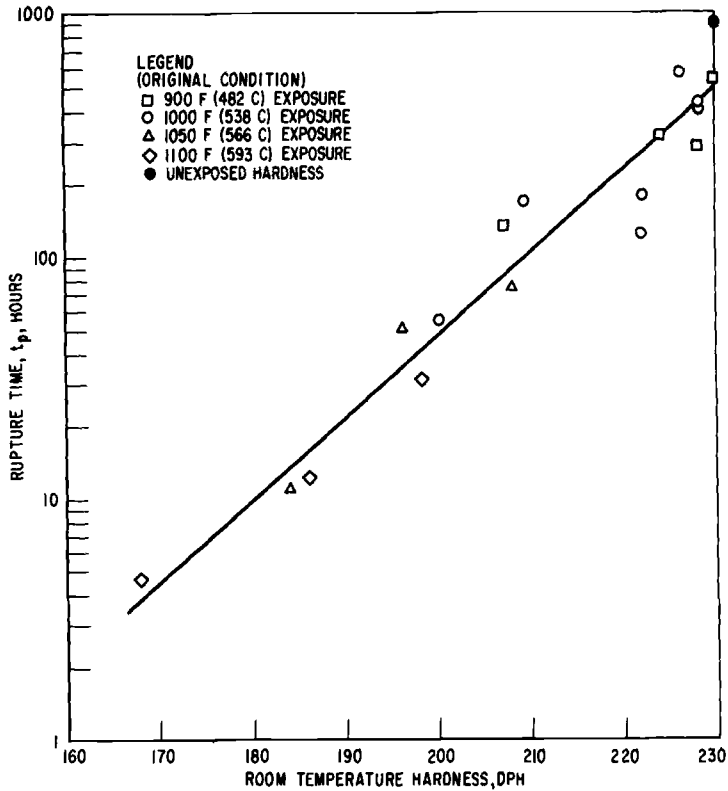


FIG. 7—Correlation between post-exposure rupture time and room temperature hardness (DPH) for Cr-Mo-V steel.

cally in Fig. 8. Here the engineer considers the accumulation of creep damage to be linear with time and that failure occurs at a constant limiting value. For any history of exposure the summation of damage indicates failure when the limiting value is reached, as in Eq 1. Temperature may be incorporated by combining with time through a simple parametric scheme. In this case the abscissa becomes a time-temperature parameter number. A similar approach has been used where the strain fractions are summed to unity. Clearly, in both methods, the limiting damage factor is assumed to be independent of the test or service conditions.

The results of this study may be organized in a scheme to show the stress and temperature dependence of creep damage with some important consequences. To illustrate the procedure, the smoothed set of original creep-to-rupture data at 1000 F shown in Fig. 9 was used. Several combinations of strain and time were taken from each curve and the values of the parameter $.01T + \log(\epsilon_0 \cdot t_0)$ were calculated. From Fig. 3, the corresponding values of t_p/t_c were then used to construct Fig. 10. Note that the ordinate is now expressed as $1 - t_p/t_c$ so that the creep damage

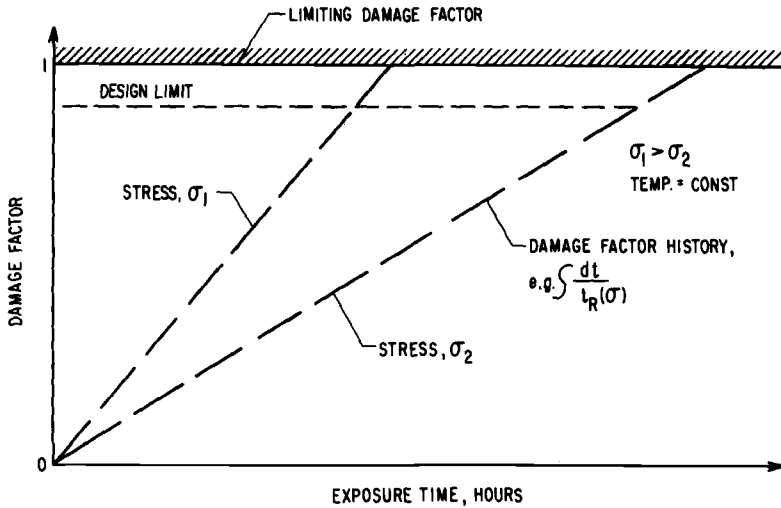


FIG. 8—Description of simple linear cumulative damage under creep conditions.

increases to a limiting value of unity as t_p approaches zero. Also included in Fig. 10 are the actual data points at 1000 F for the stresses indicated. In some cases these stresses differ from those used in the derived curves but the trend of cumulative damage is accurately maintained. Thus the master curve of Fig. 3 may be used to generate damage curves from creep curves for all test conditions yielding a parameter value in the range covered in this Figure. Curves similar to Fig. 10 were also derived from

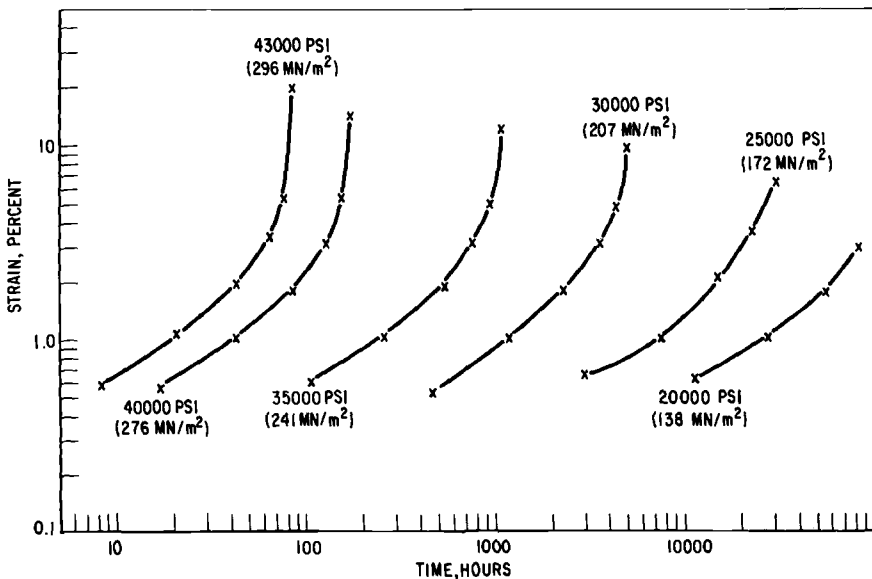


FIG. 9—Original creep data on Cr-Mo-V steel at 1000 F (538 C).

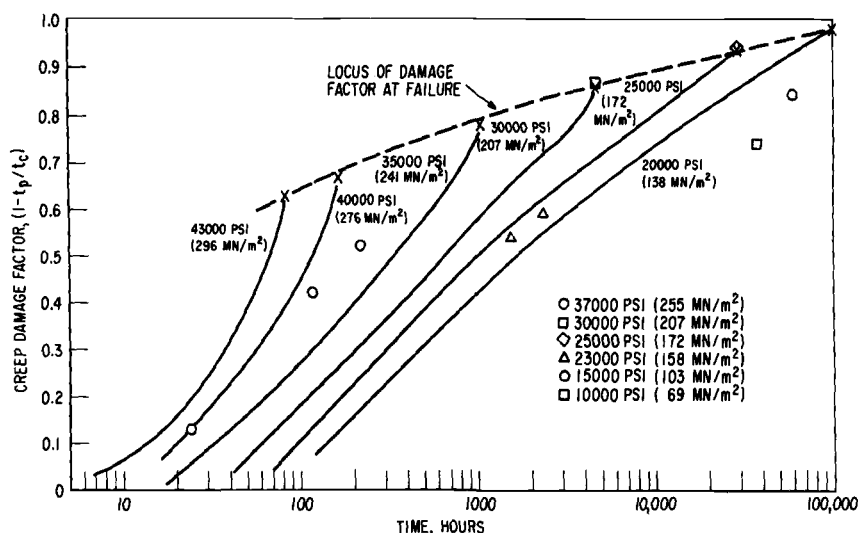


FIG. 10—Accumulation of creep damage as a function of stress for Cr-Mo-V steel at 1000 F (538 C).

available creep to rupture data at 900, 1050 and 1100 F. All these curves are shown in Fig. 11 where the abscissa of Fig. 10 has been replaced by the ratio t_0/t_R , that is, the life fraction for initial exposure.

By comparison with the idealized behavior depicted in Fig. 8, the following points may be made:

1. Creep damage accumulation in this steel is in general nonlinear with respect to the initial exposure time.
2. At high temperatures and low stresses damage is most rapid in the early stages of exposure. At low temperature and high stresses damage appears to be more rapid in the later stages of exposure. Thus there appears to be a narrow stress-temperature regime where damage is approximately linear with time, for example, at 43 000 psi and 1000 F and also between 45 000 and 50 000 psi for 900 F exposure.
3. There is no unique value of creep damage at which failure always occurs, that is, each set of tests or service conditions will produce a characteristic value of damage for failure.

Discussion

On the basis of a limited amount of rupture tests on creep exposed specimens, a correlation has been obtained with the initial exposure parameters (Fig. 3). This master curve has been used to generate damage curves for a set of creep-to-rupture data at four different temperatures. The implication of these curves in predicting cumulative damage will now be examined.

Consider the curves for 1000 F (Fig. 11). Suppose we expose a speci-

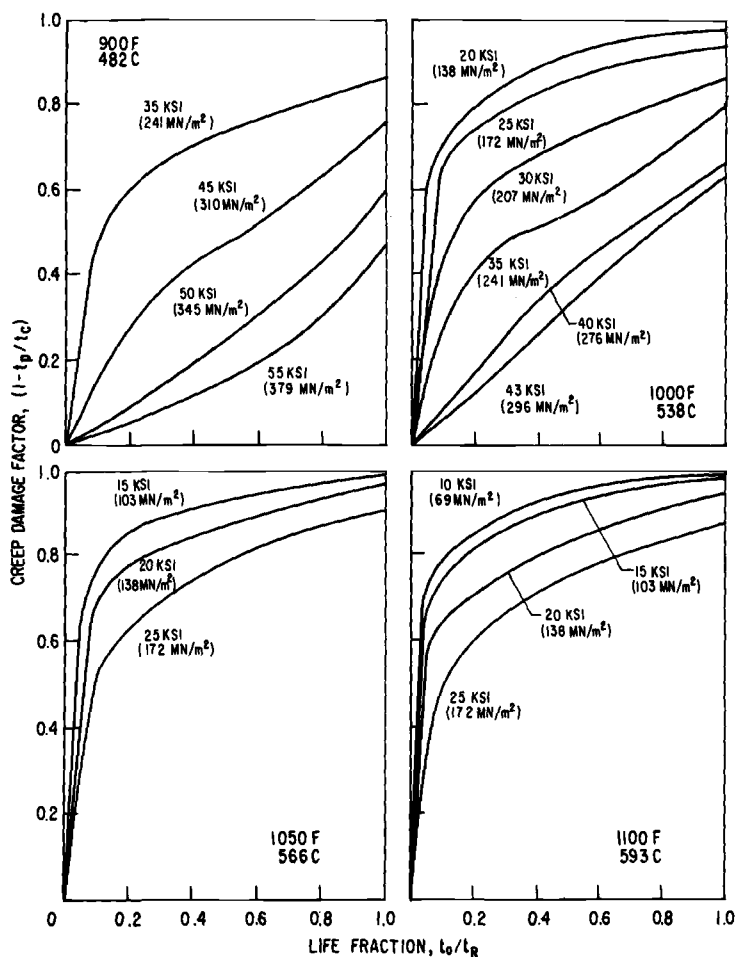


FIG. 11—Graphical description of life used up in crept specimens of Cr-Mo-V steel as a function of stress.

men at 43 000 psi for 40 h amounting to $t_o/t_r = 0.5$, the corresponding Damage Factor is 0.34. At 20 000 psi this same value is obtained at $t_o/t_r = 0.02$ or $t_o = 2000$ h. Thus 40 h at 43 000 psi produces a structural state which gives the same post-exposure life as 2000 h at 20 000 psi. This implies that a stress change from 43 000 psi to 20 000 psi at this stage would leave 0.98 of the initial life remaining at the lower stress and not 0.5 as predicted by the linear life fraction rule. Similarly, exposure at 20 000 psi for times in excess of $0.06 t_o/t_r$ would weaken the material to the extent that failure would occur immediately following a stress increase to 43 000 psi.

The question now arises as to whether one can predict failure under multiple load and temperature changes. This can be done if the post exposure rupture test adequately characterizes the structural state. Thus

two different exposures which give the same value of t_p for our post exposure test conditions should give a constant value for some other post exposure test. The hardness data indicate that this appears to be the case. In particular, the correlation with room temperature hardness in Fig. 7 indicates that similar curves to Fig. 11 could be constructed using hardness as a measure of creep damage for this alloy. Thus any horizontal cut in Fig. 11 representing a constant value of $1 - t_p/t_c$ approximates a unique measure of the structural state in terms of a mechanical property. The subsequent response should then depend only on the actual value of $1 - t_p/t_c$, and not on the detailed deformation history which led to that value.

This description has been pursued at some length because the conclusion is not obvious. For example, if in addition to softening, extensive cracking occurred, subsequent properties would be anticipated to be sensitive to the post exposure test conditions and no single test could characterize the state uniquely. However, in the Cr-Mo-V steel of this study the principal damaging process is a softening, presumably related to a progressive temperature and strain induced microstructural coarsening.

The effect of multiple changes in stress and temperature on life may now be predicted on the basis of our mechanical description of the prevailing state at any point in the deformation history. Figure 12 illustrates the procedure schematically for a test involving alternating exposures at σ_1 and σ_2 , each lasting 0.2 of the corresponding rupture life. After the second excursion at the lower stress giving a cumulative life fraction of 0.8, the damage at failure for σ_1 has been exceeded. If, however, this

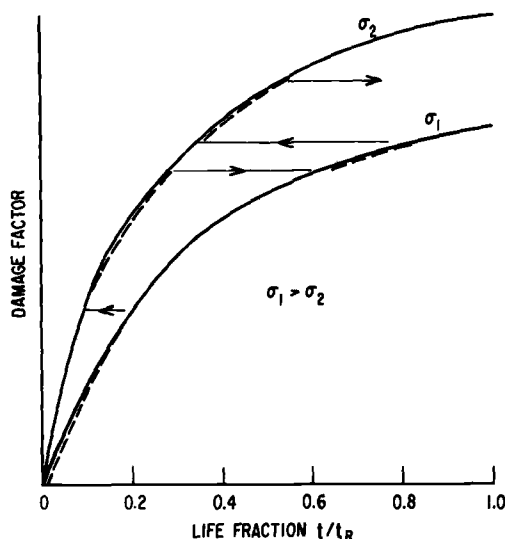


FIG. 12—Schematic illustration of damage accumulation resulting from stress changes.

final change is not made, a life fraction of about 0.45 remains at the higher stress giving a cumulative life fraction of 1.25 at failure. The curves of Fig. 10 may be used to predict life for any combination of stress and temperature in a similar manner. Conditions under which the linear life fraction rule adequately predicts life would be expected to be those involving small changes in stress and temperature, and multiple excursions. Conversely, those involving appreciable changes in test conditions, or a single excursion as in the first example in this section would show poor agreement with this rule. Based on the curves presented in Fig. 11, tests involving a single stress increase would be expected to fail at $\sum t_o/t_R < 1$ and those involving a single stress decrease would fail at $\sum t_o/t_R > 1$.

The preceding general conclusions are consistent with many experimental results reported in the literature. In the initial example cited we noted that at 1000 F, 2000 hours at 20 000 psi produces equivalent damage to 40 h at 43 000 psi. This indicates that the damage is principally a result of the deformation rather than the thermal exposure which is consistent with observations of softening after creep in a 12 percent Chromium steel by Murphy and Branch[8]. These authors found that although some softening occurred as a result of thermal exposure, the amount was small compared with that measured after creep deformation.

Gluck et al[9], working with an aluminum alloy, showed that the tensile strength dropped as much as 50 percent after creep exposure. A large portion of this drop occurred for low creep strains. Bogardus et al[10] showed that the loss in tensile strength of aluminum was greater the longer the time to reach a given creep strain (that is the lower the creep stress). For fixed creep strain and exposure time the tensile strength also decreased with increasing temperature. All these observations are consistent with the observed dependence of damage on the test conditions in the Cr-Mo-V steel.

Finally, we wish to emphasize the observed dependence of the damage at failure on the test conditions. The amount of softening calculated at failure increased with decreasing stress at constant temperature (increased test duration) and decreasing temperature at constant stress as shown in Fig. 13. A similar conclusion has previously been reached for tests on nickel[4] where the principal damaging process was one of cavitation (Type 3 damage). The common practice of assuming a unique failure criterion either in terms of cracking (for example, cavity spacing or crack length) or in terms of a post exposure mechanical property is clearly incorrect.

Conclusions

1. The post exposure rupture life of a Cr-Mo-V steel has been correlated with the initial creep exposure conditions. A master curve using

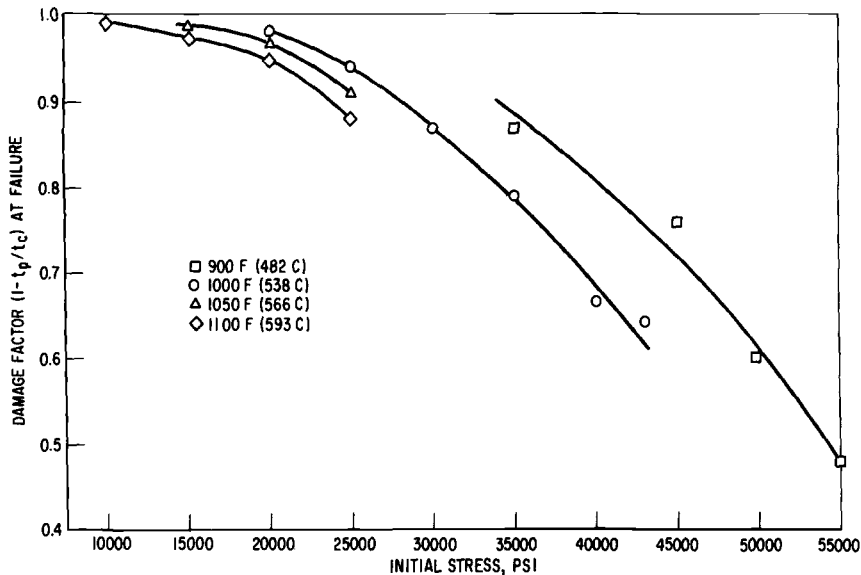


FIG. 13—Effect of temperature and stress on damage for failed specimens.

a derived parameter incorporating strain, time and temperature has been constructed.

2. Softening resulting from the creep exposure is the principal damaging process in this alloy.

3. For practical purposes the post exposure tests uniquely characterize the mechanical state characteristic of the creep damaged condition.

4. Creep damage curves have been derived from the master curve for a set of creep to rupture data covering a wide range of test conditions. These curves may be used to predict behavior under nonsteady conditions of flow and fracture. They also suggest conditions under which the linear life fraction cumulative damage rule might be expected to provide a reasonable prediction of life.

5. The damage at failure is sensitive to the test conditions. This conclusion, although consistent with other data in the literature, is contrary to the assumption normally made in life prediction procedures.

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S. A. Hoenig,¹ C. A. Savitz,¹ W. A. Ott,¹ T. A. Russel,¹ and M. T. Ali¹

Applications of Exoelectron Emission to Nondestructive Evaluation of Alloying, Crack Growth, Fatigue, Annealing, and Grinding Processes

REFERENCE: Hoenig, S. A., Savitz, C. A., Ott, W. A., Russel, T. A., and Ali, M. T., "Applications of Exoelectron Emission to Nondestructive Evaluation of Alloying, Crack Growth, Fatigue, Annealing, and Grinding Processes," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 107-125.

ABSTRACT: It has been demonstrated that exoelectron emission phenomena can be used to observe and measure alloying, crack growth, fatigue, annealing, and grinding processes. For observation of alloying, crack growth, annealing, and grinding a fixed plate or wire collector biased at +90 V is used to collect the electrons. The current is measured by a picoammeter and recorded directly via a strip chart recorder. To find the fatigued area in a specimen a heat source is scanned along the coupon and the exoelectron emission is picked up by a moving collector wire connected to a picoammeter.

These experiments are usually done in ambient air without any special control of temperature and humidity. The initiation of crack growth is observed as a sudden change in exoelectron emission level. In a fatigued but uncracked specimen a significant increase in exoelectron emission is observed when the heat source and collector move over the area where the fatigue damage is most intense.

Most of the data to date has been obtained with aluminum but we feel that this technique can be extended to other aircraft structural materials.

KEY WORDS: predictions, electron emission, nondestructive tests, alloying, crack propagation, crack initiation, fatigue (materials), annealing, grinding (comminution), stresses

The phenomenon called exoelectron emission (EEE) was first reported over 130 years ago by Moser[1]. Unfortunately, there is still a general lack of understanding of the many physical causes of EEE. However, there is enough knowledge of cause and effect relationships to consider applying EEE to specific types of nondestructive evaluation (NDE). It is suggested

¹ Electrical Engineering Dept., University of Arizona, Tucson, Ariz. 85721.

that whenever the surface of a liquid or a solid is disturbed, there is a tendency for electrons to escape into the vacuum phase. In the case of insulators or semiconductors, these electrons may come from F centers near the surface. In metals one might invoke the presence of F centers in surface oxides, but a vacancy mechanism is a more reasonable explanation of the existing experiments. It is this proposed connection between vacancies and EEE that permits application of EEE to NDE.

A number of reviews of EEE have been given recently[2-4]. At this point it is suggested that any process which disturbs a solid surface results in EEE. For example the experiments discussed in Refs 2, 3, and 4 included wire brushing, phase changes, tensile stress, annealing, sintering, alloying, X-ray excitation, ball milling, and grinding.

The major problem in studying EEE has been the fact that the currents involved are small (10^{-9} to 10^{-13} A) and the electrons have a relatively low energy (≈ 1 eV). These facts forced investigators to use Geiger tubes with all the attendant problems of gas surface interactions and non-uniform field distributions. More recently investigators have tried to use electron multipliers for detecting exoelectrons, but low energy electrons are not efficient in inducing secondary emission.

The advent of the field effect transistor and the operational amplifier has made it possible to measure current of 10^{-15} A without difficulty. This capability is a major factor in the application of EEE to practical NDE.

Experimental Studies

Exoelectron Alloying Phenomena

The formation of an alloy requires that two or more elements generate a new crystal structure with the solute atoms in either substitutional or interstitial positions. The alloying process itself requires the existence of large numbers of vacancies to permit rapid atom migration. These vacancies tend to migrate to the surface and induce exoelectron emission. The presence or absence of this emission can be directly related to the alloying process. This has been demonstrated for the alloys of tungsten or molybdenum with gold, silver, copper, and aluminum[5]. Even weak alloys such as Cu/W could be distinguished from a no-alloy case like Ag/W.

The apparatus is shown schematically in Fig. 1. Typical experiments involved heating the tungsten or molybdenum wire to 1190 C (in 5 s), holding the system at that temperature and observing the exoelectron current in a cylindrical geometry. For the alloying tests short lengths of the material to be alloyed, that is, Cu, Ag, Au, etc., were looped over the filament before heating.

Typical test results are shown in Fig. 2. As the alloying process occurs,

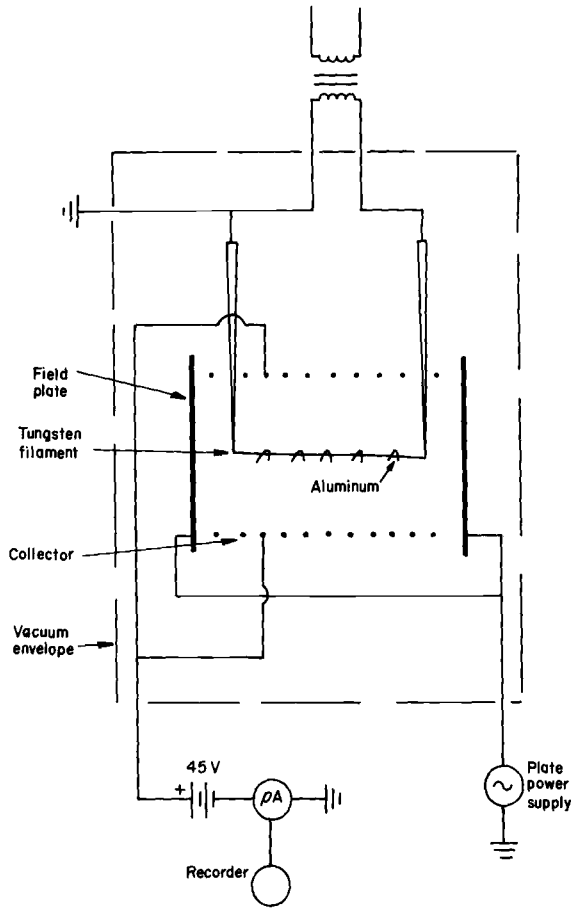


FIG. 1—Exoelectron alloying test system.

a large burst of exoelectron emission is observed. This can be contrasted with the no-alloy or bare tungsten data shown in Fig. 2.

This phenomenon can be understood in terms of the disturbance of the surface that occurs during the alloying process. It would be most interesting to investigate the relations between exoelectron emission and certain possible alloys predicted by Engels' theory[6].

Crack Growth Studies

In 1963 ASTM supported our study of the application of field emission techniques to observation of surface microcracks. Using 0.020-in. diameter wires and a cylindrical field emission microscope (FEM) it was possible to detect microcracks as small as $4\text{ }\mu\text{m}$ ($1.6 \cdot 10^{-4}$ in.). The field emission microscope is particularly suited for finding microcracks because there is no need to scan the entire surface as in optical microscopy. At

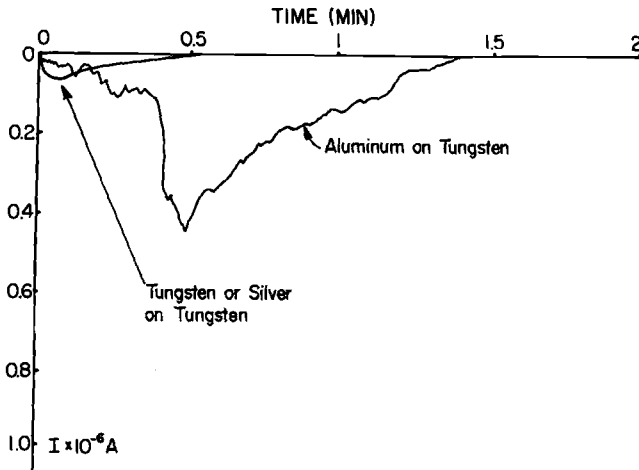


FIG. 2—Exoelectron current with and without alloying reaction.

the proper voltage only cracks and other sharp edge defects emit electrons and are imaged[1-8]. The FEM is shown in Fig. 3.

In 1967 the FEM was used to study stress corrosion cracking of uranium-molybdenum alloys. With the FEM it was possible to follow the various steps in the SCC process[9].

During these studies it was possible to pull specimen wires to failure at high (600 C) temperatures and low electric fields so that the electrons were emitted by a combined field-thermal process. The electron images were observed with a Bolex camera at 64 fps and it was noticed that emission did not cease immediately when the wire broke. The postfracture emission came primarily from the broken ends of the wire with intense

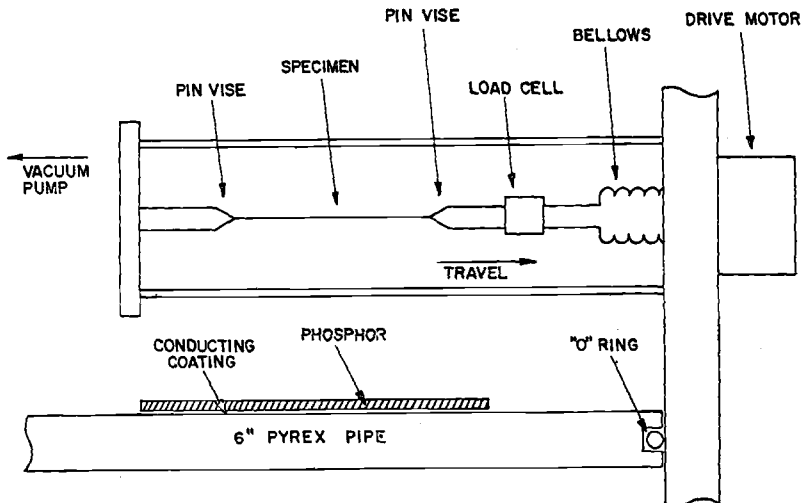


FIG. 3—Cylindrical field emission microscope for wire studies.

emission from the cup and cone areas. It was suggested that this emission was due to exoelectrons escaping from the metal as the disturbed fracture area relaxed to its equilibrium shape.

In 1970 it became possible to observe the exoelectron emission that occurs during the initiation and growth of cracks in metals. The apparatus was quite primitive; an auto jack was used to apply the stress.² The exoelectrons were monitored with the ultraviolet light and collector system to be discussed in section on fatigue evaluation.

Typical results are plotted in Fig. 4 as exoelectron current versus time as the specimen is torn apart. The appearance of cracks or tearing (under 50-power observation) and the current jumps are well correlated. The

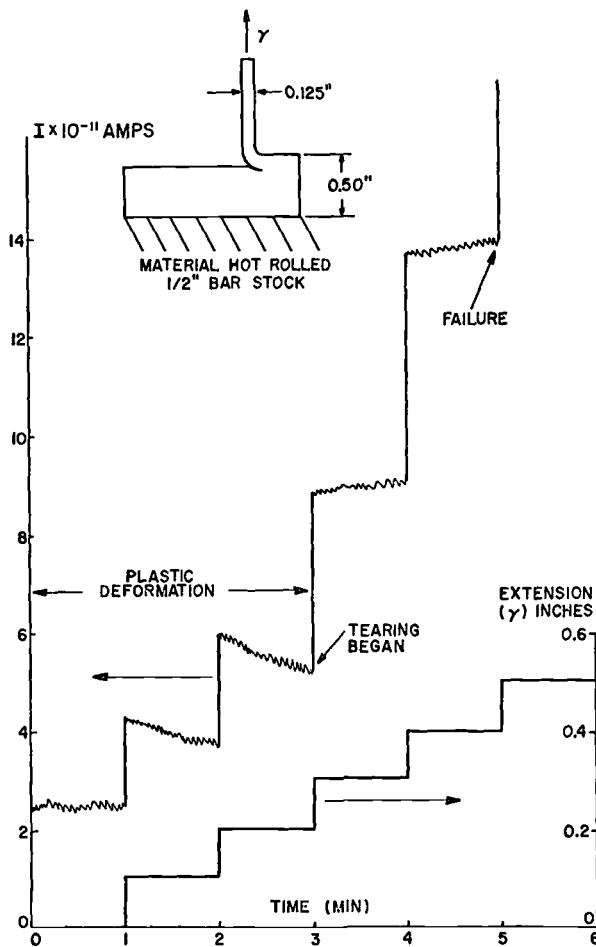


FIG. 4—Exoelectron current versus plastic deformation and crack growth.

² The auto jack system was built because of a severe shortage of testing machines at the University of Arizona.

apparatus is being redesigned for more detailed studies in which it will be possible to monitor load, deformation, and crack growth at the same time.

Fatigue Evaluation

In 1969 we began studies of the application of exoelectron emission to NDE of metal fatigue. Our program had two objectives:

1. Gain an understanding of exoelectron phenomena particularly the driving potential for emission.
2. Develop a technique for application of exoelectron emission to NDE.

It is convenient to discuss the results of objective (2) first. In this case it was known that exoelectron emission occurred during the fatigue process, but it seemed more important to develop a system for measuring the fatigue damage after rather than during the fatigue cycle. In aircraft applications it is difficult to observe and record events during flight. In any case it is most important to be able to fatigue test a critical structure before the next flight to be sure that catastrophic failure will not occur after takeoff.

It was proposed that during fatigue a complex structure of vacancies and dislocations developed near the specimen surface[10-12]. Most of these "defects" would be sessile at room temperature but upon heating a significant fraction of these defects (vacancies) might be expected to diffuse to the surface. This vacancy diffusion would in turn induce EEE, which would then be used as a measure of the fatigue level of the specimen. This application is contingent upon first, having a large enough current to measure and second, not heating the structure to the point where fundamental metallurgical changes occur.

The specimens for this study were aluminum (1100-0 and 7075-T6). The specimen shape is shown in Fig. 5. Specimens were 0.090 in. thick and were chemically polished to a 60- μ in. finish. The specimens were

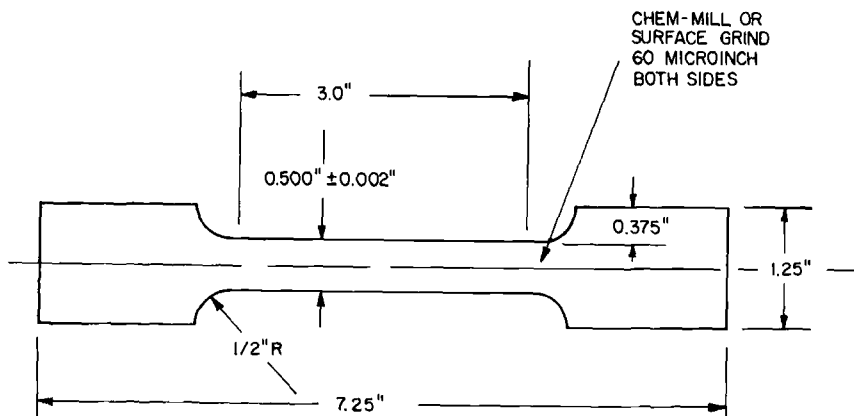


FIG. 5—Exoelectron fatigue test specimen.

fully annealed to remove fabrication stresses and then heat treated into appropriate "0" or "T6" condition.

Most of the fatiguing of the aluminum specimens was done in a resonant machine (designed and built by C. A. Savitz). Some work was done on a plate machine (Vishay Technology VSP-150) borrowed from the Metallurgy Department, University of Arizona.

To establish a mean lifetime a series of 10 specimens were run to failure. For the 1100-0 material the average number of cycles was 147 000 with a maximum excursion of ± 15 percent. The next set of specimens were run to some fraction, 0 percent, 20 percent, 40 percent, 60 percent, 80 percent of their expected life and tested for exoelectron emission.

The test apparatus and the electrical circuit is shown in Fig. 6. The ultraviolet light source seems to increase the exoelectron current by about a factor of 50, the photoelectric current is a constant background level which does not present any problem in interpreting the data. Figures 7 and 8 show the exoelectron current as a function of time for 1100-0 specimens at various fatigue levels. There is a clear, distinct relation between fatigue and exoelectron emission (at least 1100-0 aluminum). We wish to emphasize that these fatiguing and exoelectron emission studies were done in ambient air. A temperature of 90 C was reached after heating for 8 min. This is far below the temperature where metallurgical changes could occur.

Similar data on 7075-T6 aluminum is shown in Fig. 9. Here we must mention that there was more scatter in the total cycles-to-failure tests. The average lifetime (10 specimens) was 46 000 cycles ± 18 percent. In Fig. 9 we see again that the exoelectron emission can be correlated with the fatigue history of the specimen. However, we note that emission from

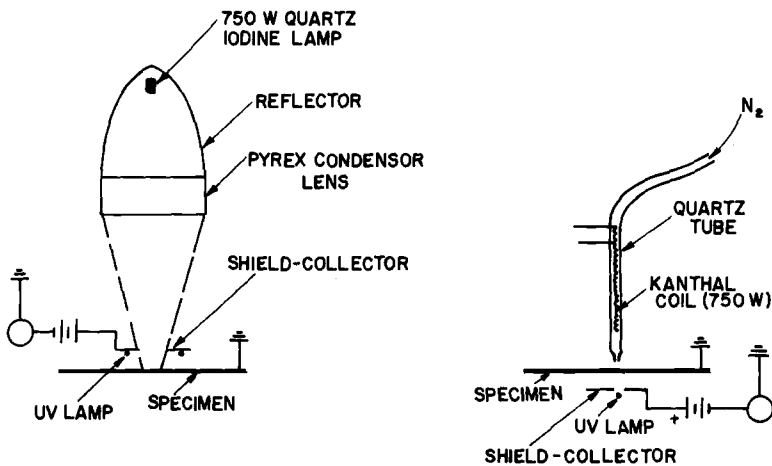


FIG. 6—Exoelectron test systems.

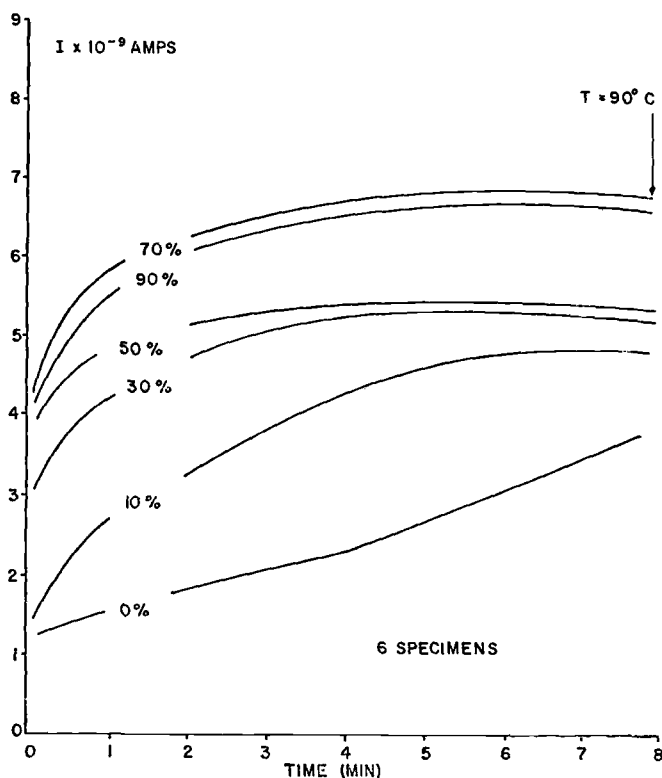


FIG. 7—Exoelectron current versus time and fatigue level (1100-0 aluminum).

the 75-percent specimen was below the 0-percent specimen until a temperature of 160 C was reached. This suggests that at high fatigue levels a higher temperature is needed to induce vacancy motion. A very similar effect has been observed with nickel wires and will be discussed below.

Notice in Fig 9 that the 25-percent and 50-percent curves overlap, this may be scatter in fatigue life itself. Alcoa data indicates that at best 7075-T6 fatigue life varies from 1000 to 14 000 cycles around a central value of 4000 cycles (80 000 psi load). Our fatigue life data is much better than that but ± 18 percent scatter could conceal a 25-percent change in fatigue level.

Tests of this type require a large number of specimens. In actual structures only a few specimens may be available and it was necessary to develop a different test technique which involved scanning the heat source and exoelectron collector along the specimen, (Fig. 6). This allows a comparison of a relatively unfatigued area with a fatigued area, on a single specimen. Figure 10 shows some recent results obtained by scanning along the specimens. It is apparent that it is possible to find the fatigued area and measure the amount of fatigue that is present. Notice in Fig. 10 that the exoelectron level, along the entire specimen, goes up

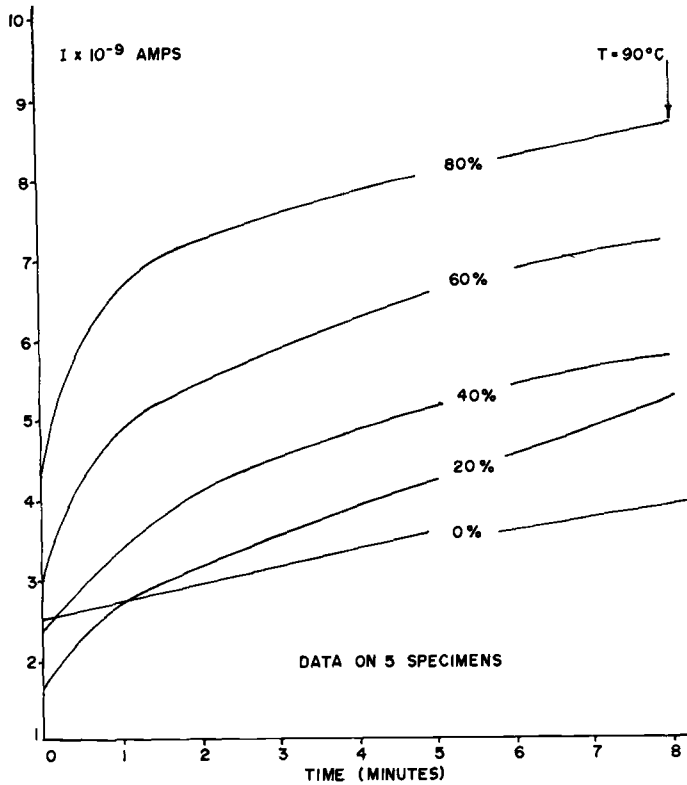


FIG. 8—Exoelectron current versus time and fatigue level (1100-0 aluminum).

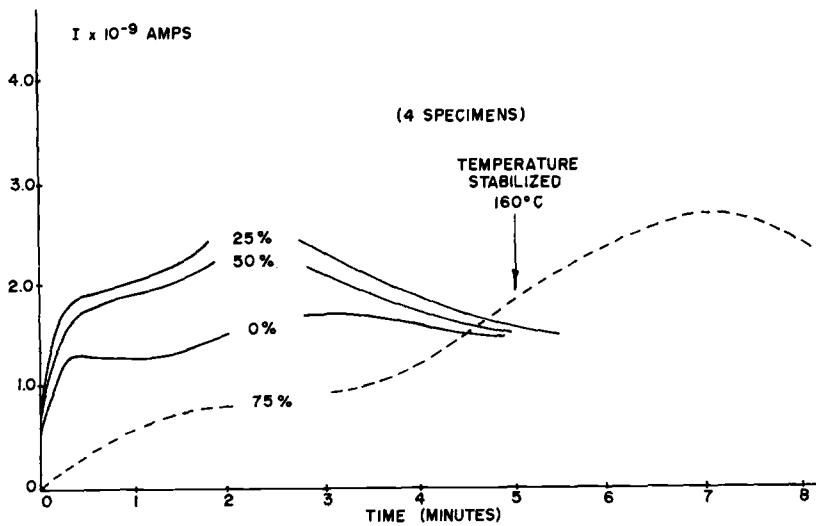


FIG. 9—Exoelectron current versus length and fatigue level (7075-T6 aluminum).

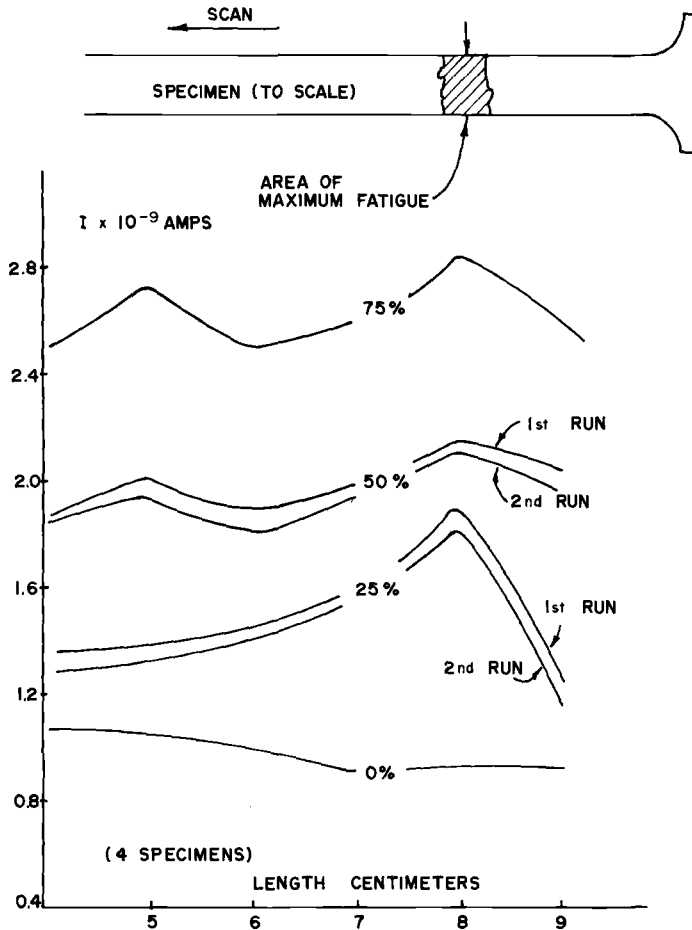


FIG. 10—Exoelectron current versus length and fatigue level (7075-T6 aluminum).

with fatigue level. This level change and the second peak observed in the higher fatigue specimens is thought to be due to the deflection cycle produced by the resonant machine.

Other work in this area has been devoted to gaining an understanding of the exoelectron phenomena. Here it has been possible to follow the work of Van Voss and Brotzen[13] where aluminum was strained in tension and the EEE was followed both during and after straining. Once again the proposed mechanism involved the generation of vacancies during plastic deformation. Subsequent vacancy diffusion was thought to release exoelectrons. Exoelectron emission persists for some time after straining ceases and this is taken as a measure of the rate at which mobile vacancies are exhausted.

Van Voss and Brotzen investigated the effect of ambient ultraviolet light on exoelectron emission. They tentatively adopted an F center model

involving a surface oxide as part of the photonelectron interaction. We suggest that when a vacancy arrives at the surface, the normal barrier potential is disturbed and high energy electrons can "leak out." If ultra-violet photons are present, the number of electrons which "leak out" is correspondingly greater.

The first experiments in this area were done by W. A. Ott for his Master's thesis in Electrical Engineering. The specimens were nickel wires 0.032 in. in diameter, necked down to a gauge section in an electro-polishing machine. The specimens were mounted in the vacuum fatigue machine shown in Fig. 11. This system can provide either a direct tensile or an oscillating torsional stress while maintaining a vacuum of 10^{-8} torr.

The specimens were annealed at 1000 C for 24 h (in vacuum) and then either strained or fatigued in vacuum at room temperature. The experiments included:

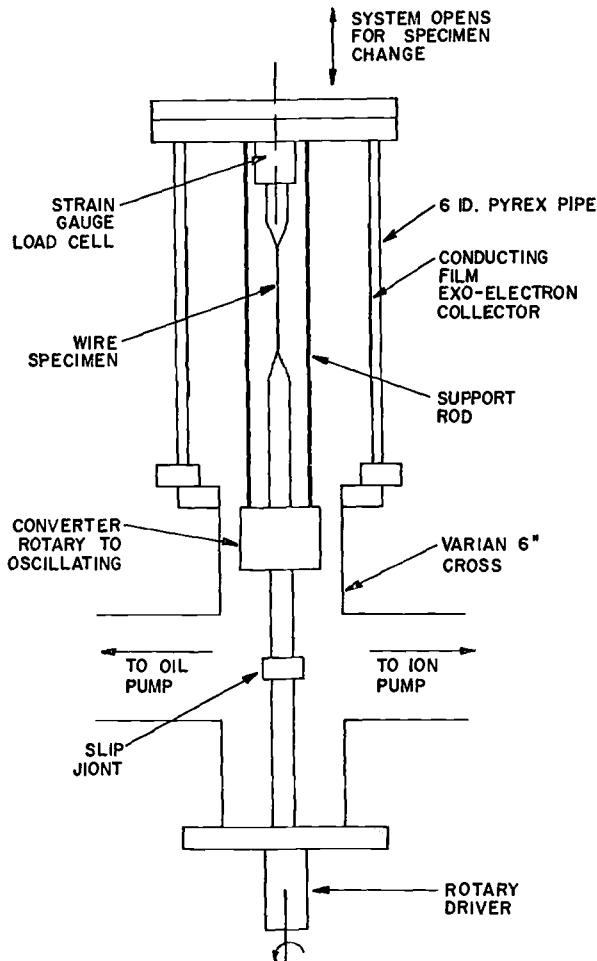


FIG. 11—Exoelectron experimental system.

1. exoelectron emission under tensile stress;
2. exoelectron emission under torsional fatigue; and
3. exoelectron emission after tensile stress or fatigue.

The results for experiment (1) above were in agreement with those of Ref 9. Exoelectron emission was observed during tension testing without any added ultraviolet source. The current increased more slowly than the tensile stress. When the load was removed the current died away over a 30-s interval. When the stress was reapplied the exoelectron current rose rapidly until fracture then decreased slowly after the wire broke.

In experiment (2) exoelectron emission at the 10^{-12} A level was observed. The pressure during fatigue was a $3 \cdot 10^{-8}$ torr. Each time the wire was twisted a burst of electrons was recorded. The currents were small and the interference from electrons emitted by the mechanical gearing made it difficult to follow changes in current level. Other investigators[14] have observed exoelectron emission during fatigue. The current first increases with each cycle and then decreases again. This may be due to work hardening, making it more difficult for dislocations to move. It is felt that measurements of exoelectron emission during fatigue are as useful as post-fatigue studies.

The studies of EEE after fatigue (3) involve heating the specimen to some fixed temperature in the vacuum system (at 10^{-8} torr) and following the EEE. The results of a typical test on several specimens fatigued to various levels are shown in Figs. 12 and 13. Again there is a clear

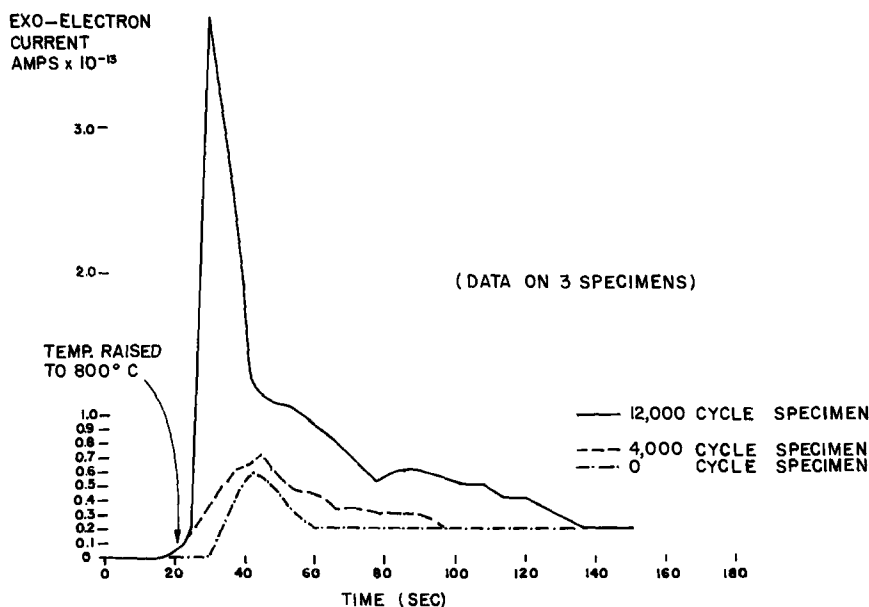


FIG. 12—Exoelectron current versus time and fatigue level (nickel 205).

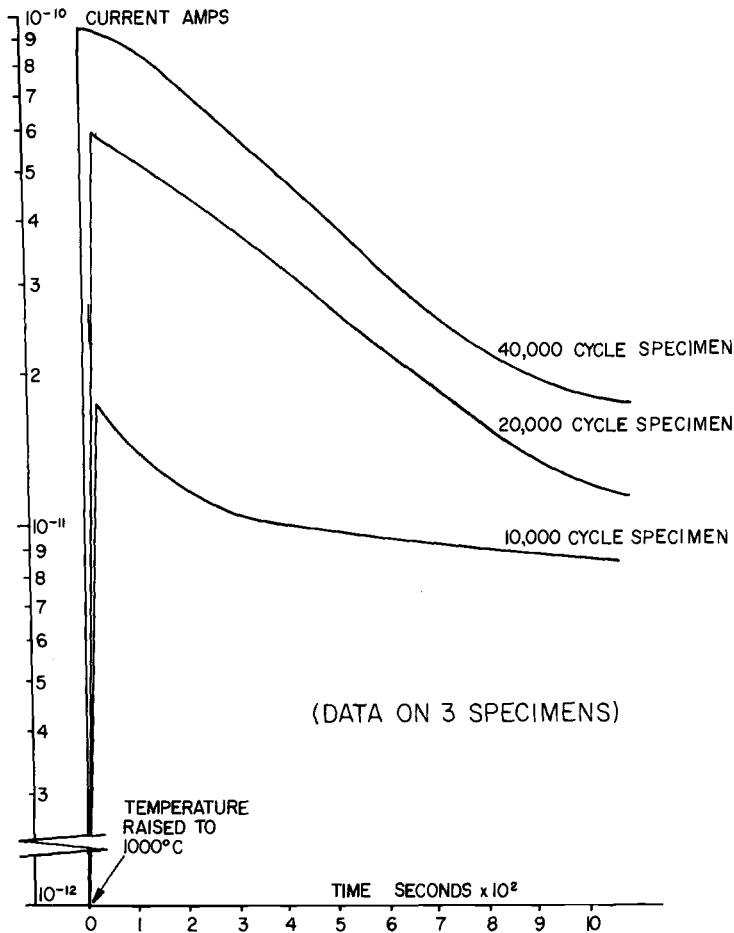


FIG. 13—Exoelectron current versus time and fatigue level (nickel 205).

relation between fatigue and post-fatigue EEE. This data was obtained without any ultraviolet light and temperatures required for significant exoelectron emission are quite high (800 to 1000 C).

It was expected that there would be a correlation between the fatigue level and the integral $\int Idt$. We began integrating the observed exoelectron current and looking at the graph $\int Idt$ versus the fatigue level of the specimen. Typical results are shown in Fig. 14, for the nickel specimens. The correlation is fair up to about the 55 percent level where the exoelectron current begins to decay. It is suggested that this represents the level at which the vacancies and dislocation clusters begin to pin one another and diffusion is impeded. This suggests that further heating, to above the 1000 C, would induce exoelectron emission at high fatigue levels. There is some evidence for this but it is not practical to heat nickel above 1000 C with the present apparatus. Because of this heating problem a decision was

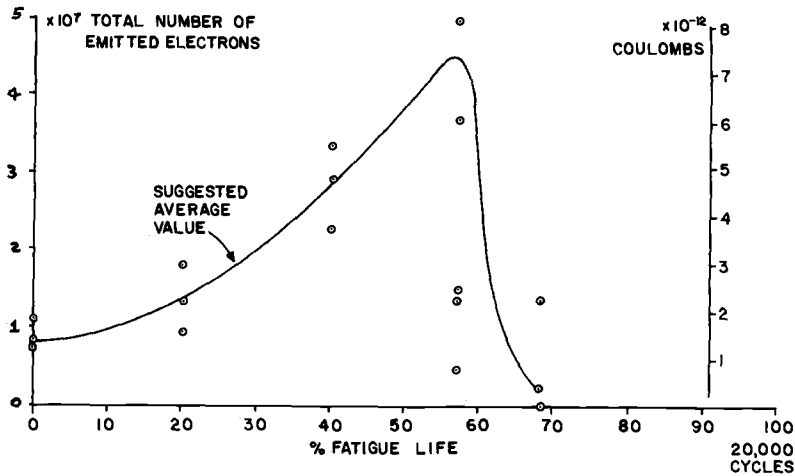


FIG. 14—Exoelectron (current) (time) product versus fatigue level (nickel 205).

made to use a quenching technique to produce a controlled number of vacancies in a metal—then heat the metal and look for exoelectrons.

Exoelectron Emission and Annealing Processes

Physicists have worked for many years on the recovery phenomenon in quenched metals. The major problem has been that the migration of vacancies was only observable in terms of secondary effects, that is, electrical resistance. The proposal that exoelectron emission might be related to vacancy migration has been reported recently by scientists in the Soviet Union[15].

It was possible to use metal wires to observe exoelectron emission, during vacancy migration, with the apparatus shown in Fig. 15. In this experiment the wire specimens are heated by a-c current via the transformer. Specimen bias was controlled by batteries. Typical experiments involved:

1. Heating the wire and then quenching it with liquid or gaseous nitrogen.

2. Annealing the wire at some fixed temperature and monitoring the exoelectron emission.

Experiments have been run on nickel and platinum, some of the nickel data is shown in Fig. 16, curve (1) shows the results of heating an unquenched wire to 680 C and holding it at that temperature for several hundred seconds. Curve (2) shows the exoelectron emission that is observed when a wire quenched from 700 to -210 C is allowed to remain at room temperature. There is a slow migration of vacancies to the surface and this produces the observed exo-electron current. Curve (3) shows the effect of heating to 680 C after a fast quench from 900 to

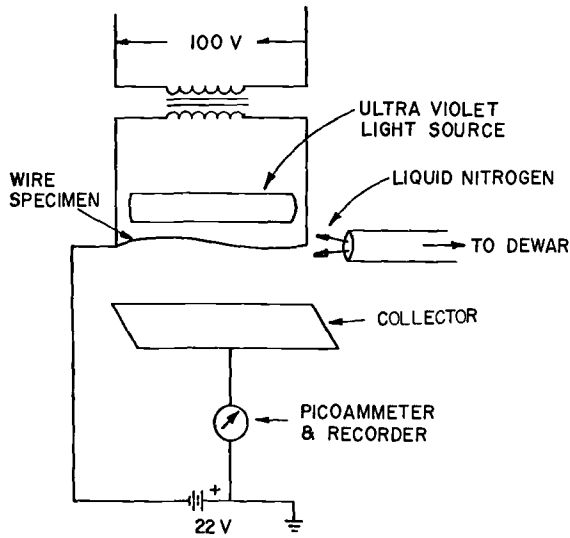


FIG. 15—Quenching and exoelectron emission system.

—210 C. The exoelectron current is much larger, and now one might speculate that there are two types of vacancies which have greatly different activation energies for migration. In contrast, curve (4) shows the effect of a relatively slow quench from 900 C and subsequent heating to 680 C. Here the exoelectron current reaches a peak and then decays as a continuously decreasing function of time. It would seem that these experiments are most significant in that they can be used to determine the relationship between vacancy concentration and exoelectron emission. A very important application of this work may exist in the use of exoelectron emission to monitor and control annealing or recovery processes.

Monitoring of Grinding and Milling Operations

Other exoelectron studies at the University of Arizona have involved the application of EEE to observation of grinding and milling processes. This was first suggested by Kramer[16] who used a modified Geiger counter to examine samples of ZnS, K₂SO₄, NaCl, gypsum after grinding. Exoelectrons were indeed observed and there was some evidence that exoelectron emission could be correlated with the grinding process. This would follow from the mechanism proposed earlier where disturbing the surface permits electrons to escape.

It seemed possible that this process might be an explanation of the fires and explosions that have occurred during grinding inflammable materials, that is, flour, paint, magnesium. In view of this data it was felt that it would be possible to monitor ore crushing processes which are of interest to Arizona's copper industry.

To test this latter suggestion a ball mill was borrowed from the

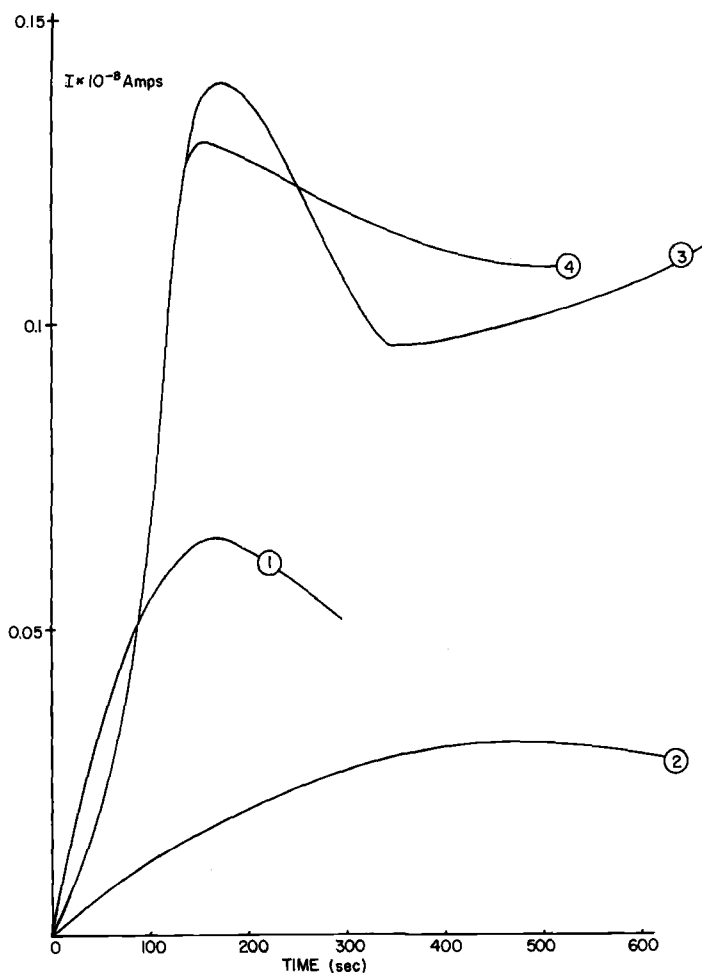


FIG. 16—Post quench exoelectron current versus time (nickel 205).

Anaconda Company Laboratories in Tucson. The ball mill and the associated electronics are shown in Fig. 17. The contact to the drum was made with brass scrubbing brushes, the center electrode was a stainless steel rod. Test batches of copper ore were run with iron balls and the iron balls were run alone for a blank experiment. The ore was milled without water, experiments with wet grinding yielded confusing results because of deposits of mud on the insulators.

Typical data is shown in Fig. 18. The observed current rises and then falls. It is suggested that the time at which the current begins to drop is the point where the ore has been ground as finely as it can be with the large balls that we used. From that point on only the iron balls were being ground which produced a small but significant residual current.

These grinding studies are continuing under NASA support. The ob-

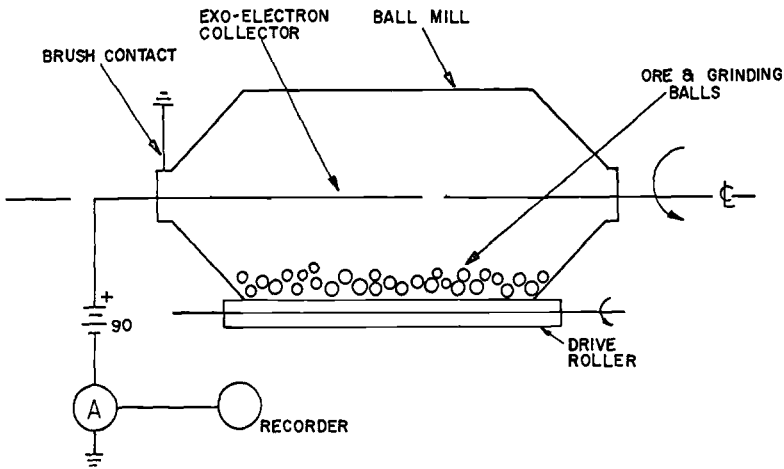


FIG. 17—Exoelectron ball mill system.

jective is to determine if the exoelectron emission observed during grinding is related to the silica content of the material being ground.

Discussion and Conclusions

It would seem that the potential of exoelectron emission for fatigue evaluation has been demonstrated, (at least in aluminum). There are still innumerable questions about how and where we can apply this technique. These studies will be part of the program during the next two years.

Another interesting application of exoelectron emission is just being

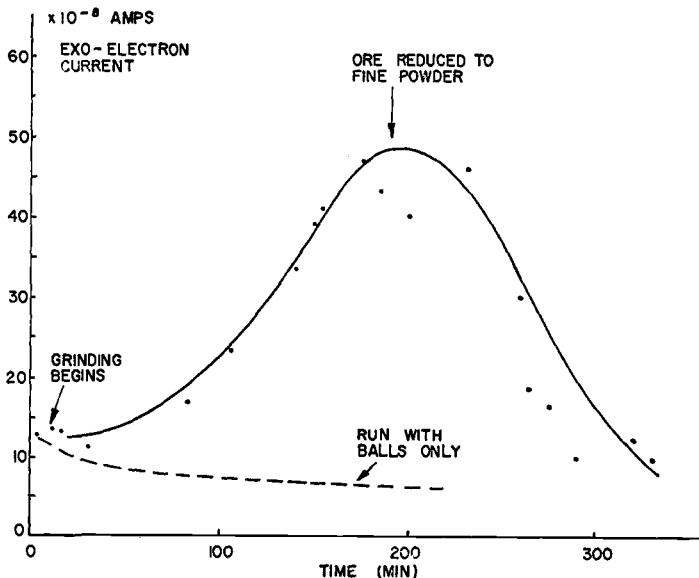


FIG. 18—Exoelectron current versus milling time (copper ore).

started in the laboratory. This relates to electron migration in thin films which is a problem in the electronic industry. When large (10^6 A/cm²) currents go through thin films of Ag or Al, the film moves by a vacancy diffusion mechanism and eventually the film conductor becomes an open circuit. A review of this phenomena has been given by Black[17]. Studies have begun to look for exoelectrons, from thin films of aluminum, under high current conditions. It is thought that this emission process was that first observed by Shenstone[18-20]; to date no other possible explanation has been given for the Shenstone effect. The ultimate objective is the development of a test technique for use during the burn-in period when integrated circuits are under load.

Acknowledgments

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We wish to thank George Martin, McDonnell-Douglas Astronautics Company, Santa Monica, California; Robert Chambers and Roy Emrick, Physics Department, University of Arizona, Tucson, Arizona; and J. Demer, Metallurgy Department, University of Arizona, Tucson, Arizona for their advice and suggestions.

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Determination of Degradation of Nylon 66 Using Differential Scanning Calorimetry

REFERENCE: Long, T. S., "Determination of Degradation of Nylon 66 Using Differential Scanning Calorimetry," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 126-142.

ABSTRACT: An analytical technique which will detect degradation in nylon 66 has been developed. This is accomplished by means of differential scanning calorimetry (DSC). The technique makes use of the double melting point phenomena to determine various degrees of degradation. This method requires a very small sample size (4 mg) and can be completed within 2 h. A discussion of other techniques and the disadvantages of each is also given. Finally, a relationship between the analytical technique and the physical properties of nylon 66 is presented.

KEY WORDS: predictions, heat measurement, scanning, nylon 66, degradation, melting points, polyamide resins, brittleness, crystallization, glass transition temperature, molecular weight

Nylon resins, especially type 66, have been used successfully for a number of years in critical mechanical applications. The material has a combination of physical properties, including excellent toughness, that makes it ideal for applications subject to impact loading. Although the performance of nylon 66 has been highly successful, some cases of brittleness have been encountered. This brittleness is usually due to degradation caused by poor process control during molding.

Degradation and the resulting reduction in part performance can be caused by: (1) excessive moisture in the molding compound; (2) excessive time or temperature or both in the melt condition during processing; or (3) excessive or improper use of reprocessed or "regrind" material.

Some analytical technique to give evidence of degradation is needed. Although many methods give this information, each has some disadvantages. Relative viscosity measurements have been used successfully, but they are time consuming and only indirectly related to physical behavior. Molecular weight determination, by gel permeation chromatography, is difficult because nylon is insoluble in many common solvents. Finally,

¹ Senior associate chemist, IBM General Systems Division, Rochester, Minn. 55901.

physical testing methods generally require large parts or samples, which do not exist in most IBM Rochester-made machines.

The criteria for an acceptable testing method includes: (1) small sample size, (2) relationship to mechanical behavior, (3) simple technique with short analysis time, and (4) reliable and repeatable results.

To establish a suitable testing method, several experimental tests were conducted. The test procedures, results, and conclusions follow. Finally, the acceptable testing procedure is described, results are described, and conclusions supporting the procedure are given.

Experimental Tests

Initially, two samples of Zytel 101² (nylon 66) were dried at 65 C for 24 h. One sample was placed in a desiccator, and the other heated at 120 C for approximately two weeks. The first analytical technique investigated was infrared spectrometry. The samples were dissolved in cresol, and a thin film was cast on a potassium bromide plate. The infrared spectra were then run on both samples (Figs. 1 and 2). Probably the most noticeable differences are the bands at 7.1 and 9.9. It would be difficult, however, to base various degrees of degradation merely on one or two bands. At this point, differential scanning calorimetry (DSC) was considered to provide the analytical technique required. Therefore, an investigation into various techniques utilizing this instrumentation was begun.

Melting Point

The two samples were run on the DSC to observe their melting peaks. Various heating rates were chosen, but each gave essentially the same

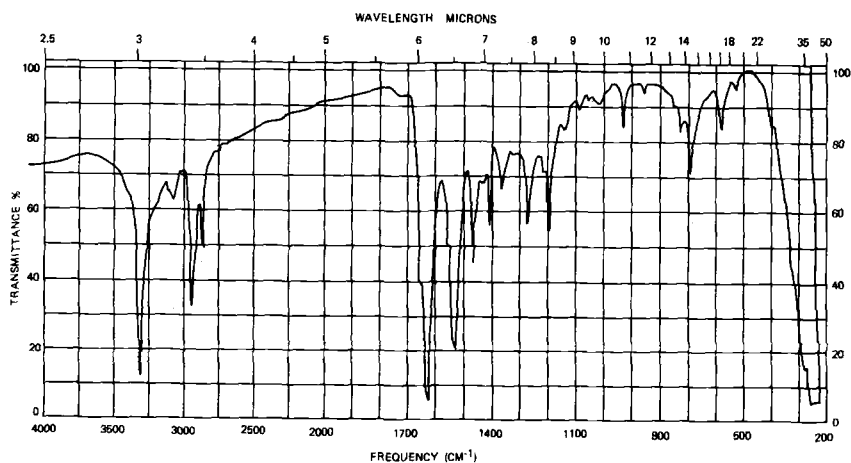


FIG. 1—Infrared spectrum of "as received" sample.

² E. I. duPont de Nemours and Co., Wilmington, Del.

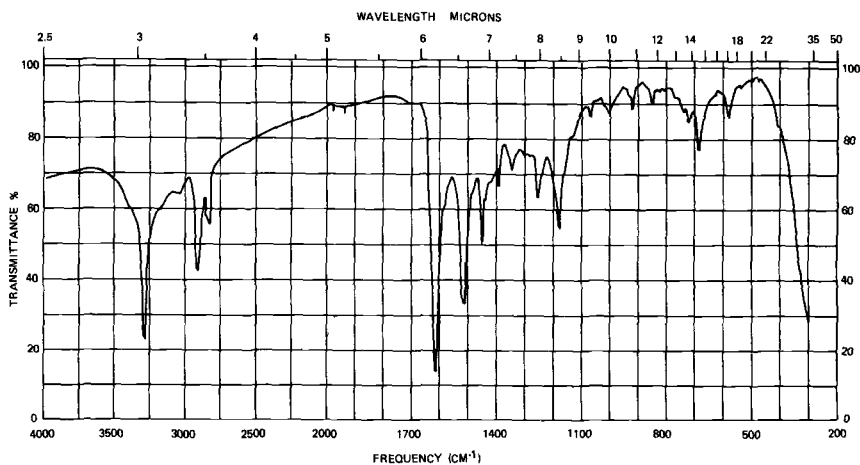


FIG. 2—Infrared spectrum of "degraded" sample.

results. Both the "as-received" and the "degraded" samples showed melting peaks at 269 C. Typical melting peaks of the two samples are shown in Figs. 3 and 4.

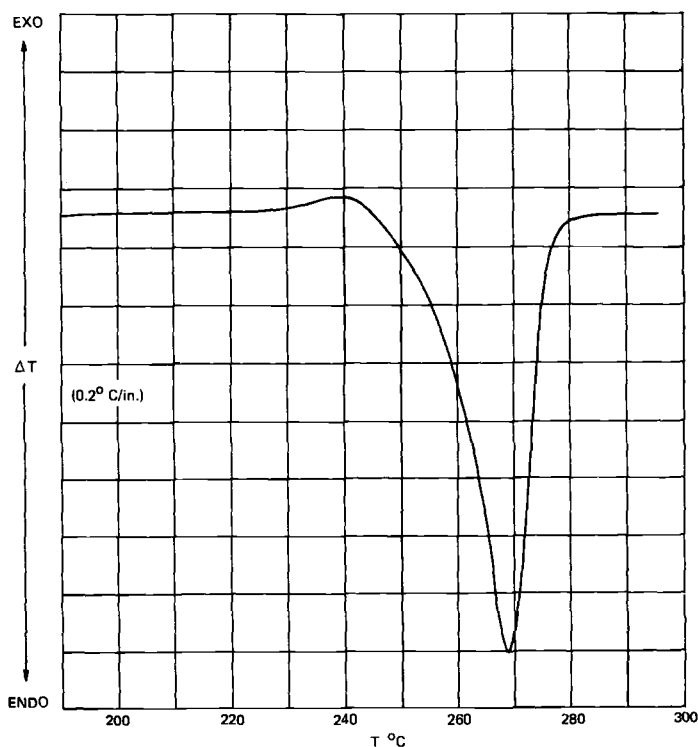


FIG. 3—DSC melting peak of "as received" sample.

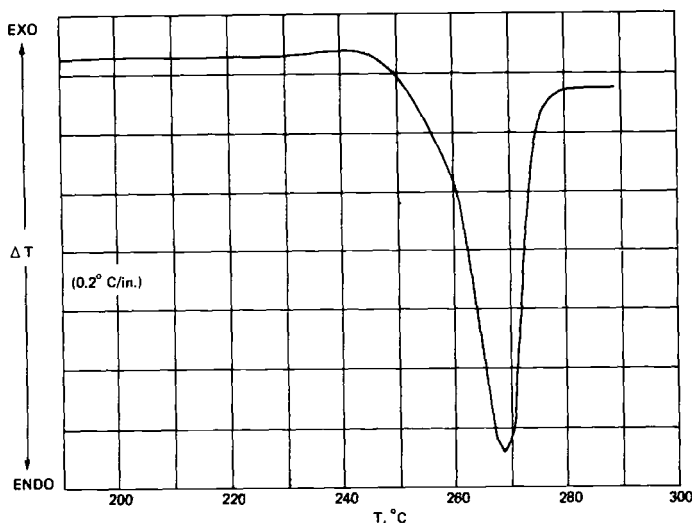


FIG. 4—DSC melting peak of "degraded" sample.

Heat of Fusion

Ke[1] indicates a direct relationship between heat of fusion and polymer crystallinity. We felt that differences might be observed in the percent crystallinity due to degradation. Therefore, the degraded and the as received heats of fusion were determined, with the degraded samples having an average of 15.43 cal/g and the as-received samples having 15.75 cal/g. No apparent differences exist between the two samples.

To further relate crystallinity to degradation, the area of the crystallization thermogram was measured. According to Ke[2], consecutive measurement of the melting and crystallization thermograms gives information on how the crystalline nature of the polymer may vary according to crystallization conditions. Consequently, a ratio of the crystalline area to the melting area was obtained by dividing one by the other. The ratios were essentially the same for both samples, so further investigation of this type was ended.

Glass Transition Temperature

Keavney and Eberlin[3] show a relationship between glass transition temperatures and molecular weight. Generally, as glass transition temperatures increase, the molecular weight tends to increase. Also, the important properties of nylon, such as impact strength, are all dependent on its high molecular weight. As the molecular weight decreases, the physical properties decrease. Unfortunately, the differential in the glass transition temperatures of the two samples was so slight that meaningful results could not be obtained (Figs. 5 and 6).

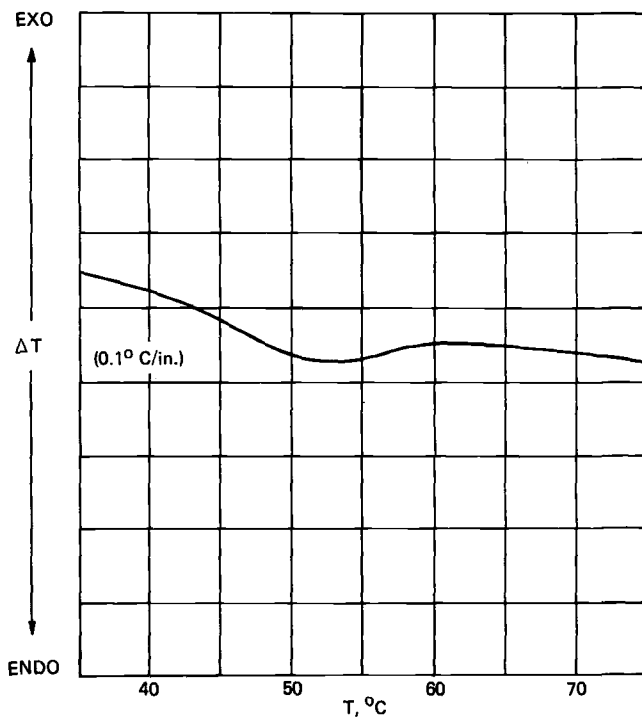


FIG. 5—DSC (T_g) of "as-received" sample.

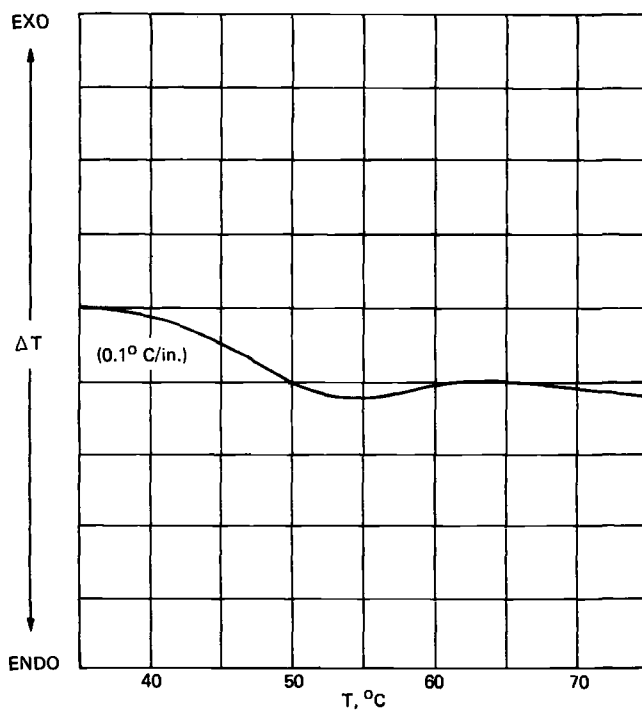


FIG. 6—DSC (T_g) of "degraded" sample.

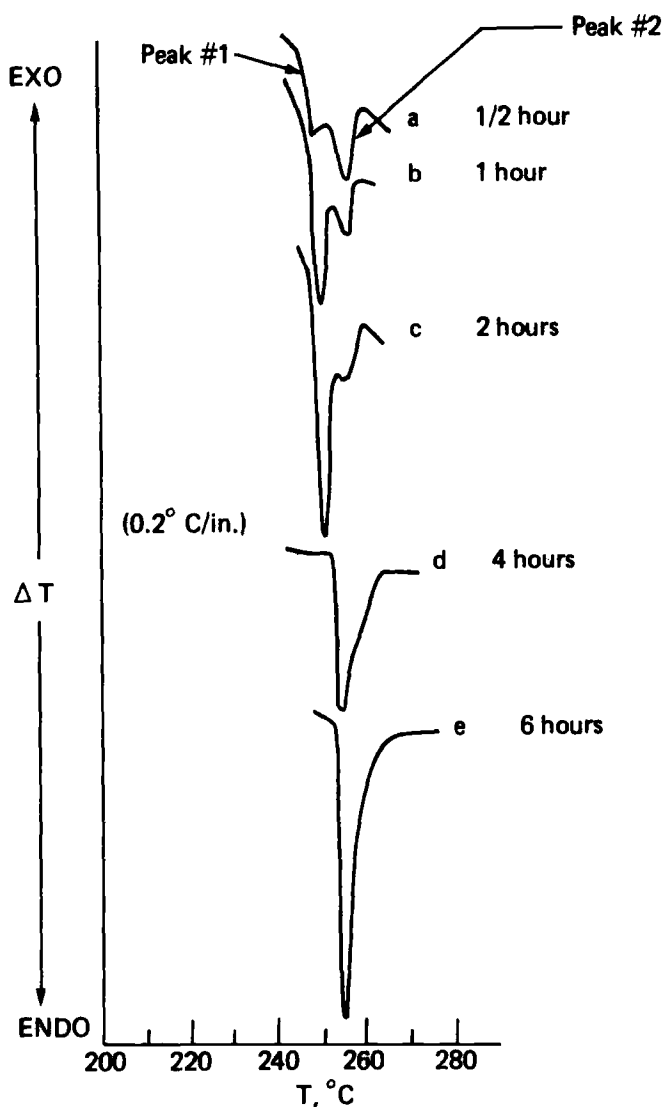


FIG. 7—DSC melting traces for samples of ground 66 nylon polymers annealed at 242°C for various times [6].

Crystallization

According to the literature[4], a higher molecular weight fraction will experience a greater degree of supercooling and, therefore, crystallize in a shorter time than a lower molecular fraction. Knowing that molecular weight has an effect on physical properties, this principle was investigated.

The time interval in each sample was measured from the onset of crystallization (245°C) until the crystallization rate reached a maximum. Three as-received and three degraded samples were timed using this tech-

nique. The supercooling in the as-received samples required 178, 182, and 155 s. The time intervals for the degraded samples took 165, 170, and 174 s. Apparently, this technique is not reproducible nor dramatic enough to give reliable information on degradation.

Two Melting Peaks

Ordinarily, a nylon 66 material exhibits only one melting point regardless of degradation. This was plainly shown earlier in Figs. 3 and 4. In recent work by Hybart and Platt[5], two endothermic peaks have been observed in the melting range. The formation of the two peaks could be observed only if the sample was taken through and above the melting point, cooled slowly, and then remelted. The samples were cooled to the onset of crystallization, or approximately 245 C, before remelting. Peak 1 was approximately 5 C lower than Peak 2.

The authors also showed that, depending upon the time the samples were held at this onset of crystallization, the peak height could be altered

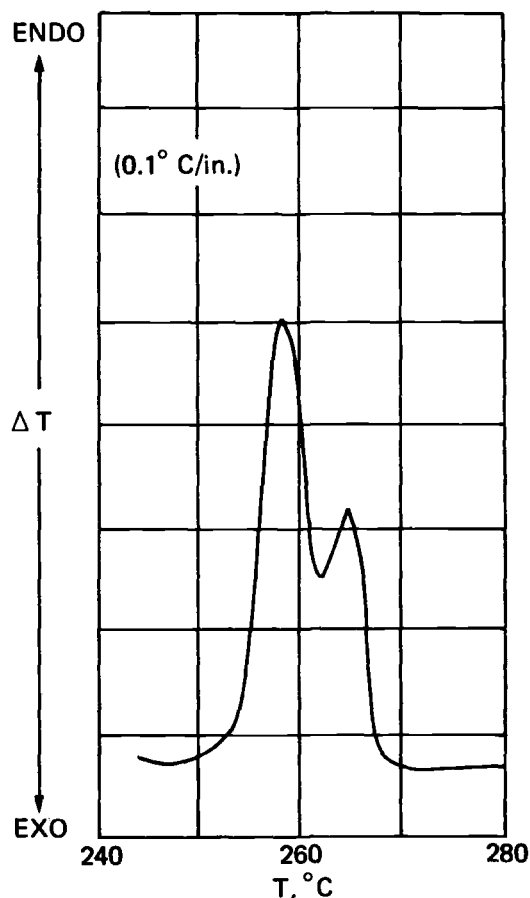


FIG. 8—DSC two melting points annealed for 1 h.

(Fig. 7). Note that a 30-min isothermal crystallization results in Peak 2 being greater in height than Peak 1. However, after 60 min, Peak 1 becomes greater than Peak 2. After eight h, Peak 2 disappears entirely. Our studies verified this, but not as dramatically (Figs. 8 and 9). We did not get a reversal in the peaks, but did obtain a reduction in peak height. However, Figs. 8 and 9 are thermograms of nonvirgin samples which have been subjected to considerable heat before analysis, due to molding conditions. Hybart and Platt considered the first peak formed during secondary crystallization or an annealing process, rather than primary crystallization.

Bell and Dumbleton[7] have attempted to explain this phenomenon of double melting. They proposed that one endotherm (Peak 2) is caused by the melting of folded chain crystals, while the other (Peak 1) is due to the melting of less perfect bundle crystals. Peak 1, therefore, contains crystals in which chains within the crystal are extended.

Starting with information from the literature, a procedure for attaining

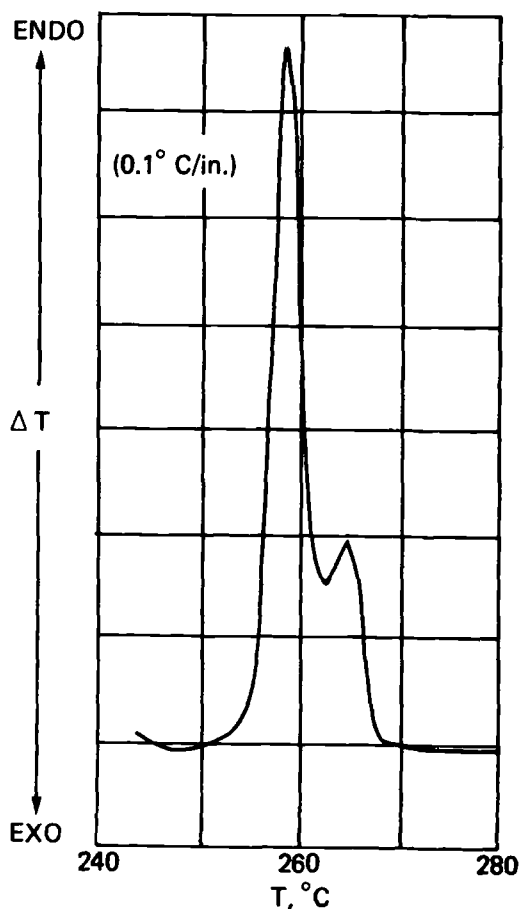


FIG. 9—DSC two melting points annealed for 2 h.

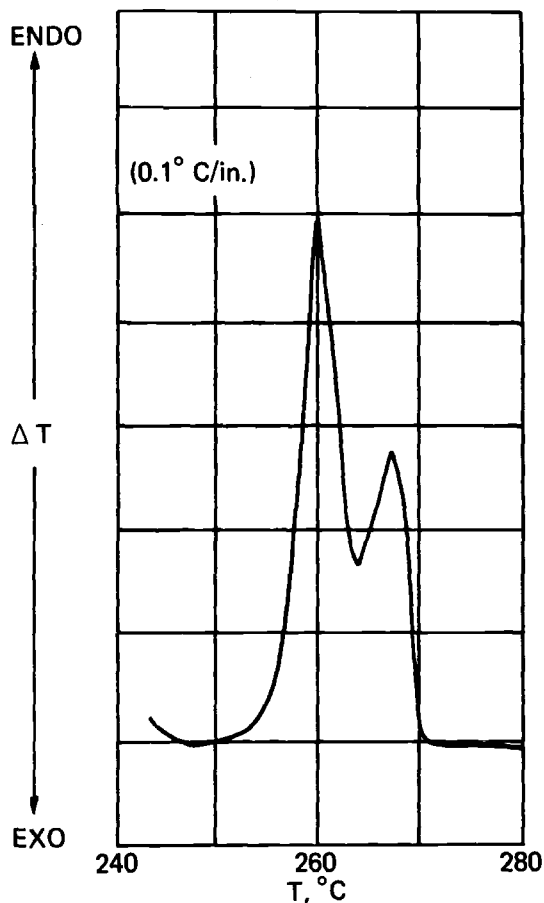


FIG. 10—DSC two melting points "as received" sample.

the two melting points was developed. The significance of this method is shown in Figs. 10 and 11. There is an obvious reversal in the peak heights of the as-received and the degraded samples. However, the total areas of the peaks are virtually the same in both samples (approximately 1 in.²). Three thermograms of each sample were run, and Peak 1 in the as received sample was nearly twice the height of Peak 2 in all cases. In the degraded sample, however, Peak 1 was slightly smaller in height than Peak 2. Therefore, it appeared that significant differences in degradation could be detected using this method.

Procedure

Sample Preparation

Tension test and Izod type impact test samples were molded from virgin nylon 66 material (Zytel 101).² The molding conditions were standard production procedures using facilities normally found in a pro-

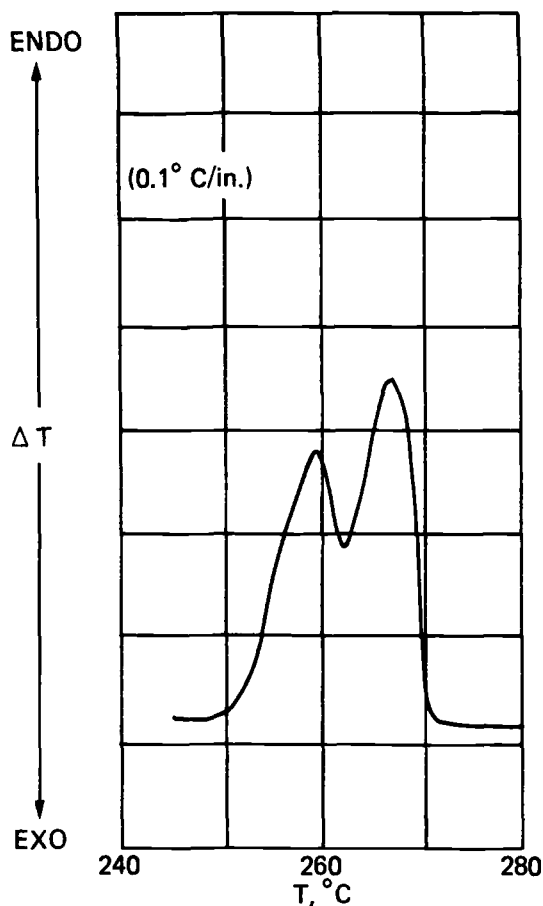


FIG. 11—DSC two melting points "degraded" sample.

duction type environment. The tension test samples were molded to conform to ASTM D-638 Type I Test for Tensile Properties of Plastics, with a $\frac{1}{8}$ -in. cross section. The Izod type impact test samples were molded as ASTM D-256 Tests for Impact Resistance of Plastics and Electrical Insulating Materials, also with a $\frac{1}{8}$ -in. cross section.

The test samples were then dried for 24 h at 65 C, after which ten samples were immediately placed in a desiccator and labeled "0 days degradation." Sixty additional samples were degraded at 120 C for periods of 1, 2, 3, 4, 7, and 10 days. The samples were dried and degraded in an air atmosphere. After each period of degradation, the samples were also placed in a desiccator.

Technique

The samples were analyzed in a duPont 900 differential thermal analyzer² with a DSC cell attachment. This is the method used:

1. Select sample size of 4.0 to 4.5 mg;
2. Heat to 240 C at approximately 30 C/min while purging the cell with 100 ml/min of dry nitrogen;
3. Slow heat rate to 5 C/min and heat to 280 C;
4. Hold temperature at 280 C for 10 min;
5. Cool to 245 C at 5 C/min (245 C is the onset of crystallization);
6. Hold temperature at 245 C for precisely 1 h;
7. Remelt at 5 C/min through the two melting peaks to approximately 280 C;
8. Measure the height of Peaks 1 and 2 from the baseline in mm; and
9. Establish a ratio by dividing Peak 1 by Peak 2.

The above technique has some important considerations which should be followed to attain the desired results. The small sample size was first sug-

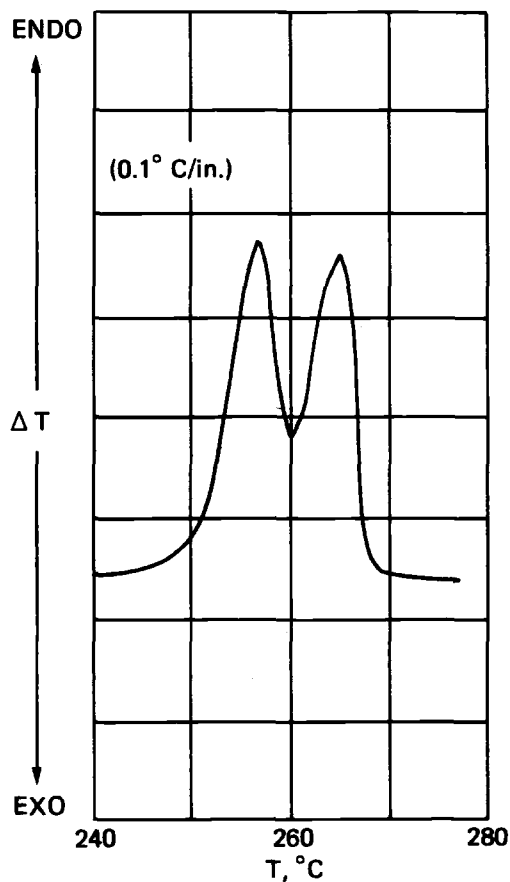


FIG. 12—DSC two melting points annealed at 240 C.

gested by Hybart and Platt[5], where better resolution of the two peaks was obtained using a small sample rather than a large sample.

The cooling temperature should not go lower than the onset of crystallization, or 245 C in this case. If this temperature drops to 240 C, peak height will be greatly affected. This was observed many times, even though the temperatures were lowered to only 243 C. Therefore, the temperature for crystallization (245 C) must be accurately measured. A peak height discrepancy is shown in Figs. 12 and 13. The sample in Fig. 12 was crystallized at 240 C and the sample in Fig. 13 at 245 C. Peak 1 is obviously larger in one than in the other.

The cooling rate is also critical. Bell and Dumbleton[7] have shown that the peak height can be varied simply by slow or rapid cooling of the sample after melting. Generally, the height of Peak 1 can be increased by slow cooling from the melt, whereas Peak 2 is increased by rapid cooling from the melt.

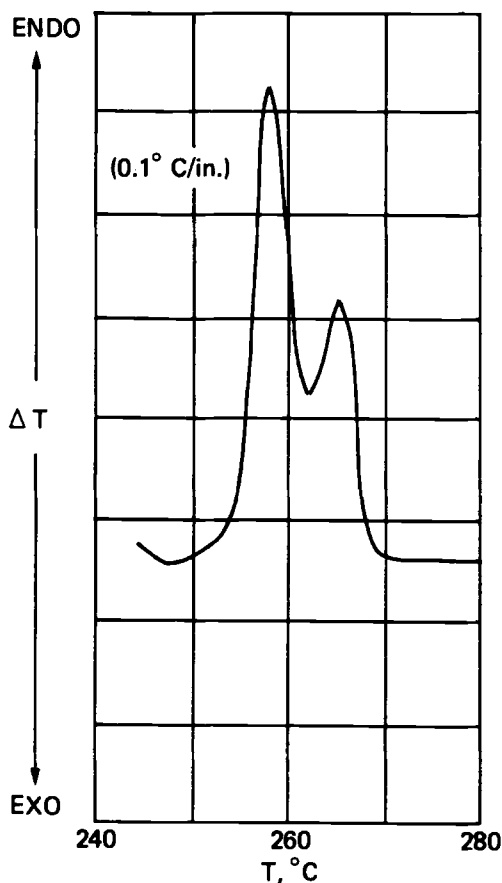


FIG. 13—DSC two melting points annealed at 245 C.

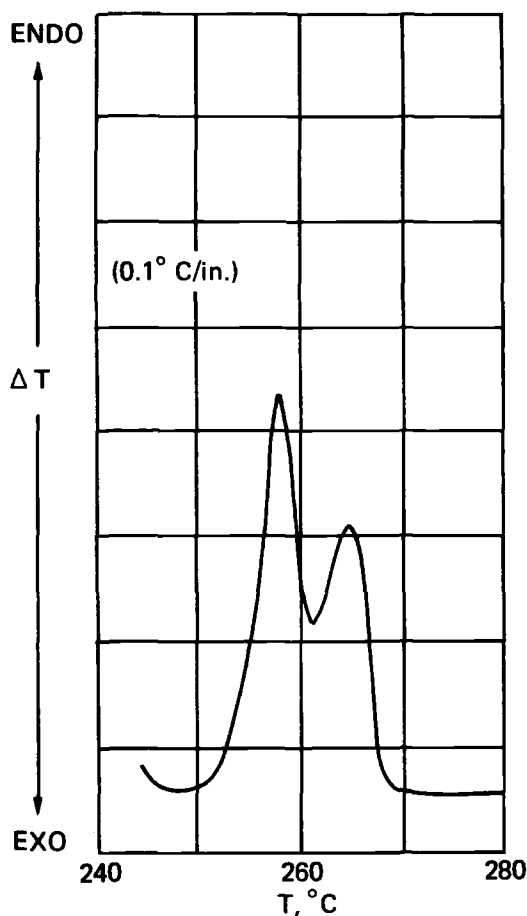


FIG. 14—DSC two melting points annealed for 60 min.

The 1-h crystallization time must be accurate to within 15 s. Figure 7 shows the effect of varying crystallization time on peak height. Even an extra 5 min of crystallization can effect the peaks, especially Peak 1 (Figs. 14 and 15). A 1-h crystallization time was selected because of additional work by Hybart and Platt[5], which showed that heating for periods of more than 15 min, at the onset of crystallization, helped produce better resolution of the two peaks. Evidently, the first peak becomes sharper and increases in size as the crystallization time is increased.

Various methods of interpreting the data were considered. As indicated earlier, the total areas of the two melting peaks were nearly identical in the as-received and the degraded samples. Another possibility was to measure the areas of the individual peaks. However, this can be very time consuming when using a planimeter or a triangulation method. Also, because the two peaks do not resolve to complete separation, it is difficult to measure the area of each individually. Finally, some kind of peak

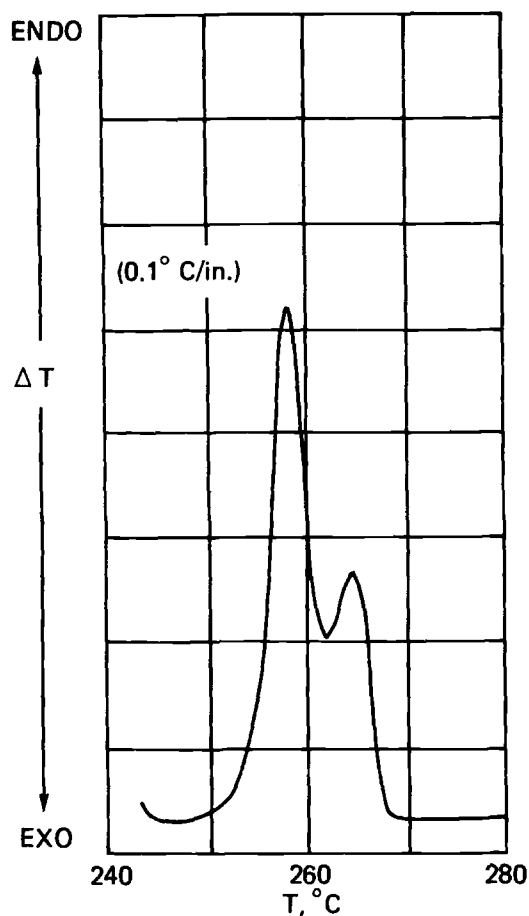


FIG. 15—DSC two melting points annealed for 65 min.

height measurement with sample weights was considered. Eventually, the best results were obtained by simply measuring and dividing Peaks 1 and 2 to establish a ratio.

Results and Discussion

We intended to relate an analytical technique to the mechanical behavior. Consequently, the samples were subjected to various physical tests. Only the elongation and the Izod impact tests are included in the results, because they showed the greatest change and are more closely related to the toughness of nylon 66. The results are given in Table 1 along with the ratios as determined by the two melting point technique described earlier.

The data shown in Table 1 are based on an average of 10 samples. Therefore, 70 measurements were taken to compile the table. Occasionally, the physical measurements were erratic, especially the elongation values. Although this is not too unusual, there are possible explanations:

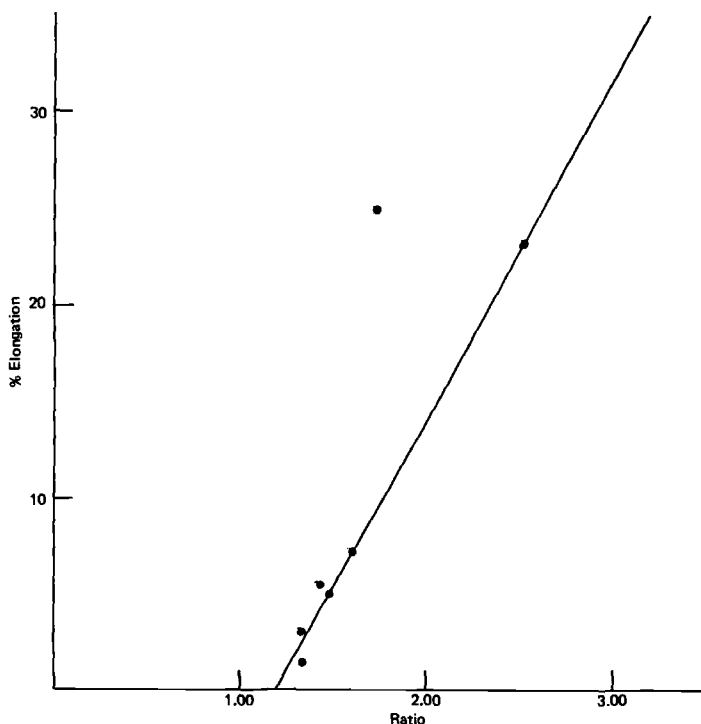
TABLE 1—*Analytical and physical data on nylon 66.*

Days of Exposure @ 250 F	Ratio	Elongation (%) ^a (ASTM D 638)	Impact Strength Izod (ASTM D 256) ft/lb/in of notch
0	2.53	23.2	0.33
1	1.73	24.9	0.20
2	1.61	7.2	0.05
3	1.43	5.4	0.05
4	1.48	4.9	0.02
7	1.33	3.0	...
10	1.34	1.4	...
Virgin resin	2.49		

^a Properties determined in a dry "as molded" condition.

(1) stress concentration points due to flash removal; (2) possible differences in processing, such as inadequate temperature control of the mold; or (3) introduction of small gel particles which never melted.

Occasional small differences were also detected on similar samples in the two melting point technique. This was probably caused by uneven degradation of the total sample. The most reliable results were obtained when the 4-mg sample was taken near the outer surface. This conclusion is supported by evidence found in a study conducted by the E. I. duPont de Nemours Company[8], which states that surface area of a nylon 66 material will undergo more severe degradation than the interior.

FIG. 16—*Relationship of melting ratio versus percent elongation.*

The relationship between the melting ratio and the physical data is given in Figs. 16 and 17. The data show degradation up to a certain point, after which testing methods seem to lose their sensitivity.

Conclusions

The method presented should be useful as a guideline to determine degradation in nylon 66. Although the absolute ratio might vary somewhat from lot to lot, it can generally be compared directly. Certainly a brittle part will exhibit a lower numerical ratio than a tough part, regardless of the lot number.

According to the data, as shown in Table 1, a ratio which falls below 1.70 will probably have poor physical properties. Consequently, this numerical ratio might be used as a criteria to determine good and bad parts.

The important contribution of this paper is the originality of the technique and its possible application to serious problems of degradation.

Acknowledgments

I wish to thank K. J. Mack, Chemistry and Polymers Department, IBM, Rochester, Minn., for his valuable suggestions and contributions

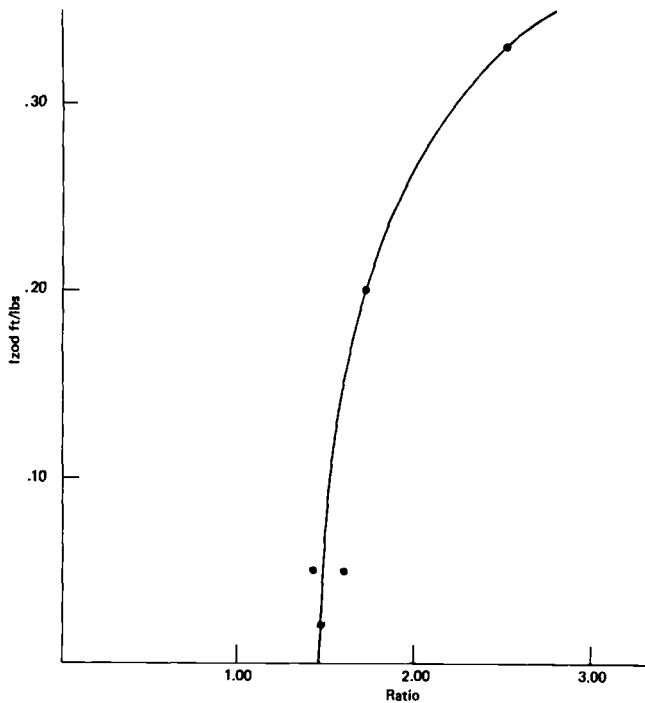


FIG. 17—Relationship of melting ratio versus Izod impact test.

during this study, and C. H. Balster, Chemistry and Polymers Department, IBM, Rochester, Minn. for conducting the numerous physical tests.

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The Early Detection of Fatigue Damage by Exoelectron Emission and Acoustic Emission

REFERENCE: Moore, J. F. and Tsang, S., "The Early Detection of Fatigue Damage by Exoelectron Emission and Acoustic Emission," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 143-157.

ABSTRACT: The paper describes the development of nondestructive test methods capable of determining the extent of fatigue damage and providing a means of predicting the future safe life of aerospace materials and structures. The fatigue process of 1100-0 aluminum was studied by means of exoelectron emission and acoustic emission measurements. The exoelectron observations were made in both vacuum and air. The exoelectron emission measurement in vacuum was accomplished by counting the emission events for selected time intervals. The measurements in air were performed using a specially developed current-measuring system with a resolution of 10^{-15} A. The acoustic emission measurement in air was based on the number of acoustic events occurring in selected time intervals during fatigue. The exoelectron emission curves exhibited an early rapid rise and fall of emission intensity, and again a rapid rise near the end of the test. The initial emission decrease coincided with metallurgically observed changes of surface slipbands. There was also a trend indicative of a stress-independent relationship between the change of exoelectron emission current and the percentage of life after a selected number of fatigue cycles. Analysis of the acoustic emission data showed a marked increase in emission rate in less than 50 percent of the fatigue life of the material.

KEY WORDS: predictions, nondestructive tests, emission, acoustics, electron emission, fatigue (materials), damage, elastic waves, stresses

The phenomena of fatigue of metallic materials and the detection of material damage due to these phenomena have always been high on the list of material problems. The problem of fatigue is generally attacked by two approaches: basic studies through physics of metals, and phenomenological observations in materials of metallurgical events occurring during fatigue tests. Knowledge derived from both approaches un-

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¹ Los Angeles Division, North American Rockwell Corporation, Los Angeles, Calif. 90009.

doubtedly contributes much to the understanding of the problem, but it is still of little use toward developing nondestructive testing (NDT) methods for the detection of fatigue damage.

Fatigue damage is assessed in different ways. One criterion is based on the residual static strength after cracks grow to a certain size; another is based on the prediction of accumulations of damage when a specimen or a structure has been subjected to a number of cycles or load spectrum. Both criteria have important practical aspects in engineering applications, however, NDT should consider the detection of fatigue damage before a macrocrack develops.

There is evidence in the literature[1] that when all the slipbands appearing on the metal surface during fatigue were removed by electro-polish, the specimen life was lengthened in the subsequent test. In the case of annealed copper, the specimen was still unbroken after $2\frac{1}{4}$ times the normal fatigue life. Evidently, the fatigue damage was removed when all the slipbands on the metal surface disappeared.

It is known that most physical properties of metallic materials undergo changes when the materials are subjected to plastic deformation. Some of these properties should also be influenced by fatigue stressing. It is then possible to measure these changes nondestructively and to utilize the measured quantities as indicators for the extent of fatigue damage. The physical properties which have been investigated include magnetic permeability[2,3], magnetic hysteresis, and eddy current losses[4], electrical resistance or impedance[5,6], light reflection[7], internal friction[8-10], and propagation of ultrasonic waves[11-13].

Of all the changes in these physical properties relating to fatigue damage before formation of macrocracks, it seems that only the propagation of ultrasonic waves, particularly surface waves, may be the suitable candidate for developing a proper nondestructive test method. Use of other properties is questionable. This collaborates the view offered by Barrois[14] who found that changes in damping, modulus of elasticity, magnetic permeability, and electrical resistivity do not provide a measure of fatigue damage.

Exoelectron emission and acoustic emission—not yet widely known in the nondestructive test fields—were considered to have greater potential for assessing fatigue damage. Exoelectron emission is the emission of electrons from both metals and nonmetals which are or have been subjected to treatments such as abrasion; mechanical deformation; irradiation with ultraviolet light, X-rays, or ultrasonic waves; electron bombardment; electric gas discharge; chemical interaction between the material surface and the ambient gas; phase transformation upon heating or cooling; etc. Since exoelectrons are emitted from the surface of a material, the emission must be a surface phenomenon. Since fatigue cracks usually start at or near the surface, both fatigue damage and exoelectron emission belong to

surface phenomena, and the change of one may be used as a reference for the other.

The phenomenon of exoelectron emission in the fatigue process has been reported by several authors[15–20]. The intensity of the emission generally increases as the test proceeds. After it reaches a saturated value, it slowly decreases until specimen failure. A rapid rise of intensity shortly before failure has also been observed. This rise is attributed to the formation of fatigue cracks, but the progressively changing emission events before the final rise can be analyzed to yield some indications of early fatigue damage.

Acoustic emission is the propagation of a spontaneously generated elastic wave created by discrete movements within the volume of the material being stressed. The generation of elastic wave is associated with such plastic deformation processes as twinning, slip, and microcrack formation. This phenomenon also includes the propagation of released strain energy upon fracturing. Acoustic emission has been used to study the fracture process[21] and to monitor crack growth in a precracked specimen so that its fatigue life can be predicted by combined fracture mechanics and acoustic emission techniques[22].

Since slip movement in plastically deformed metallic materials can cause acoustic emission, and since slipbands invariably appear on a metal surface in the fatigue process, there should be a certain relationship between acoustic emission and metallurgical change by fatigue. It is therefore possible to detect fatigue damage due to this structural change by determining the acoustic emission characteristics of originally unflawed materials.

Experimental Apparatus and Material

Fatigue Test Unit

Tensile fatigue (zero to maximum load) tests were performed using a hydraulic loading system capable of operation at 120 cpm and fitted with a chamber for vacuum and air testing.

Exoelectron Emission

The intensity of exoelectron emission from a specimen in fatigue testing was measured both in vacuum and in air. A Bendix 4039 electron multiplier, connected as shown in Fig. 1, was used in the vacuum test. The semiconductor type multiplier is helical and has a horn of 10-mm diameter at one end. The horn is positioned at the middle portion of the specimen at a spacing of approximately $\frac{3}{8}$ in. and is enclosed in a shielded box equipped with an aperture. The shielded box is charged at the same positive potential as the bias voltage to the electron multiplier so that only the electrons emitted from the specimen surface can enter the electron multiplier. The multiplier has a gain of about 10^8 with an application of

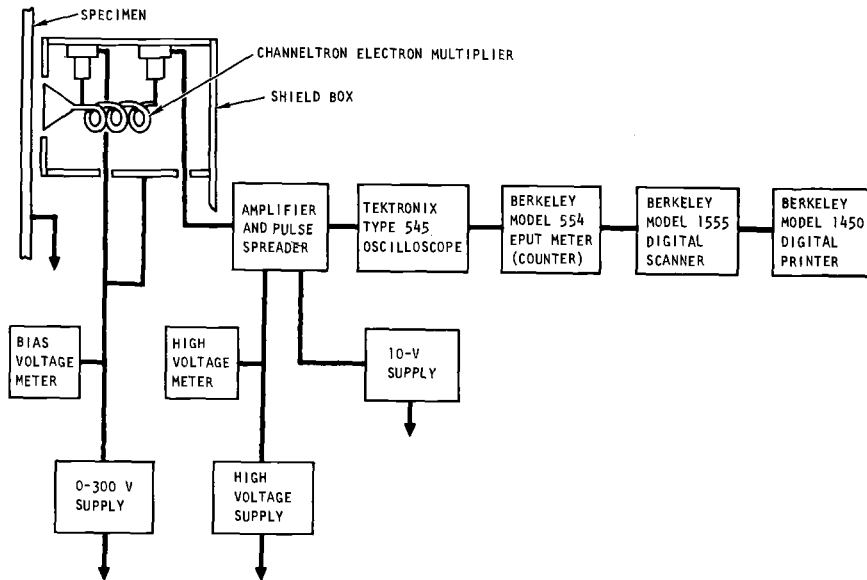


FIG. 1—Block diagram for measuring exoelectron emission in vacuum.

3000 V(dc). The pulses from the electron multiplier are amplified and stretched before they are counted. The number of counts in either 1- or 10-s periods is recorded with a digital tape printer.

The exoelectron emission current was directly measured in the air tests, as opposed to the events per unit time measurement used in the vacuum tests. Since the current is generally very small, it is necessary to stimulate the emission to increase the current level to more readily measured levels. A special measuring system (Fig. 2) was devised to measure current as small as 10^{-15} A. The current entering the collector plate from the specimen is measured as a function of the amplitude of the average voltage applied to the feedback resistor. The voltage is converted to digital readings by an Electro Instruments Model 8409 electronic digital voltmeter, and is recorded using a Hewlett-Packard Model 561A digital tape recorder. The time interval between printing (10^1 , 10^2 , 10^3 , or 10^4 cycles) is controlled by the cycling of the hydraulic system of the fatigue test unit. The exoelectron emission was stimulated with a shortwave ultraviolet pencil lamp, shielded in a metal box. The current collector plate consists of a fine stainless steel screen having about 80-percent open area. The ultraviolet light leaves the box through a $\frac{1}{2}$ -in.-diameter hole, passes through the screen, and projects on the center portion of the test specimen.

Acoustic Emission

The acoustic emission from the specimen during fatigue test in air was detected using a Dunegan D-1408 acoustic emission transducer. This

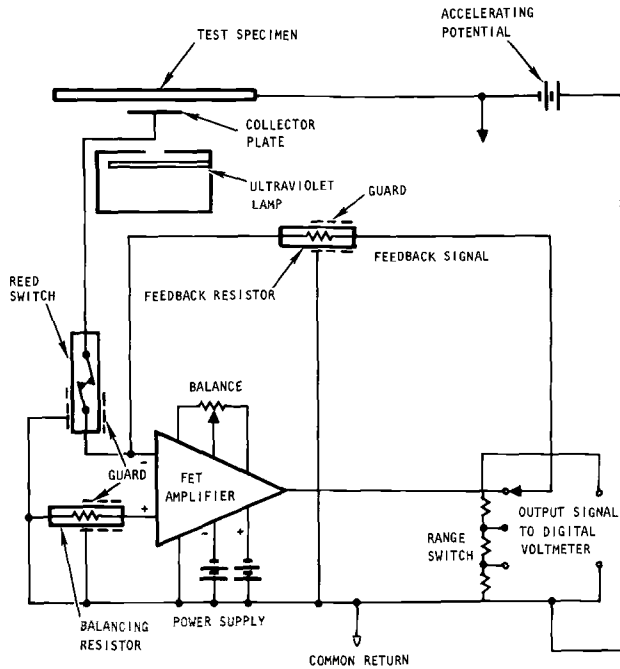


FIG. 2—Schematic diagram of the system for measuring exoelectron emission in air.

transducer is electrically shielded, has a differential output, and a narrow bandwidth with resonance frequency of approximately 210 kHz. The transducer is held in position on the specimen by a spring clip and is acoustically coupled to the specimen surface with vacuum grease.

The electrical output signal from the transducer was amplified by an oscilloscope preamplifier and passed to the input terminals of a digital counter. Any signal with amplitude greater than an arbitrary trigger threshold was counted and printed on paper tape as the number of counts in a given number of fatigue cycles. A bandpass filter in the preamplifier was used to select only those frequency components between 100 kHz and 1 MHz.

Material and Specimen

1100-0 aluminum was selected for the initial fatigue studies using exoelectron and acoustic emissions methods. The mechanical properties of this material are: ultimate tensile strength, 12 760 psi; tensile yield strength, 6530 psi; elongation at failure, 31.80 percent in 2 in.

Standard fatigue test specimens 7 in. long and 1 to 1½ in. wide were cut from the 0.09-in.-thick sheet. The sides of the test section were machined to 6-in. radius giving specimen widths of either 0.42 or 0.50 in. After machining, the specimen surface was polished chemically or electrolytically. Two solutions were used in chemical polishing: (1) a solu-

tion similar to Keller's etch for aluminum but containing twice the usual amount of hydrofluoric acid; and (2) a chemical milling solution containing sodium hydroxide (20 oz/gal), sodium gluconate (2 percent), and sodium sulfide (2 percent) heated to a temperature not exceeding 170 F. Electropolishing was performed in either of two baths. The methanol (75 percent) and nitric acid (25 percent) bath was kept below 50 F; the phosphoric acid (60 percent) and sulfuric acid (40 percent) bath was heated to between 160 and 170 F.

Experimental Methods and Results

Exoelectron Emission

Test Method—All fatigue tests involving exoelectron emission measurements were conducted at room temperature and 120 cpm. The applied stress varied between 10 000 and 11 450 psi. The vacuum tests were performed in the 10^{-5} -torr range. The tests in air were performed with the specimen and instrumentation transducer enclosed in the glass chamber to lessen air movement around the specimen and to prevent drifting of the emission current. The specimen was irradiated with short-wave ultraviolet light for about 18 h to insure that an equilibrium state of electron emission existed after specimen preparation and mounting.

Results—Figure 3 shows the exoelectron emission curve for specimens subjected to fatigue stress at 11 450 psi under ambient light and vacuum conditions. The emission intensity, expressed in number of counts per 10 s, increased shortly after the fatigue cycling was initiated. The emission appeared to take place in bursts, as evidenced by the numerous peaks in the curve. The strongest intensity occurred at about 800 cycles followed by a decreasing trend until the level decreased to the initial level at 4.5×10^4 cycles. Upon further cycling, the measured intensity remained at the initial level until about 6.7×10^5 cycles. The intensity rose again, and fatigue crack was observed. Finally, the intensity rose rapidly as the crack traversed the specimen, causing complete fracture of the specimen.

Three additional stresses of 11 000, 10 540, and 10 000 psi were applied to various specimens to investigate the effect of fatigue stress level on exoelectron emission. The emission characteristics from specimens loaded at 11 000 psi exhibited similar variations as previously noted. However, the emission was generally weaker, and the intensity decreased to the initial level earlier. Tests were stopped before specimen failure. The exoelectron emission from specimens subjected to lower stress (10 540 and 10 000 psi) appeared to increase initially at approximately 100 cycles and exhibited a very low intensity level. The highest emission level was about $2\frac{1}{2}$ times the initial level and occurred for a shorter period.

Some sources have claimed that electrons do not emit from deformed aluminum in darkness and that light must be provided as a stimulator. One specimen was therefore subjected to 11 450 psi in vacuum and under

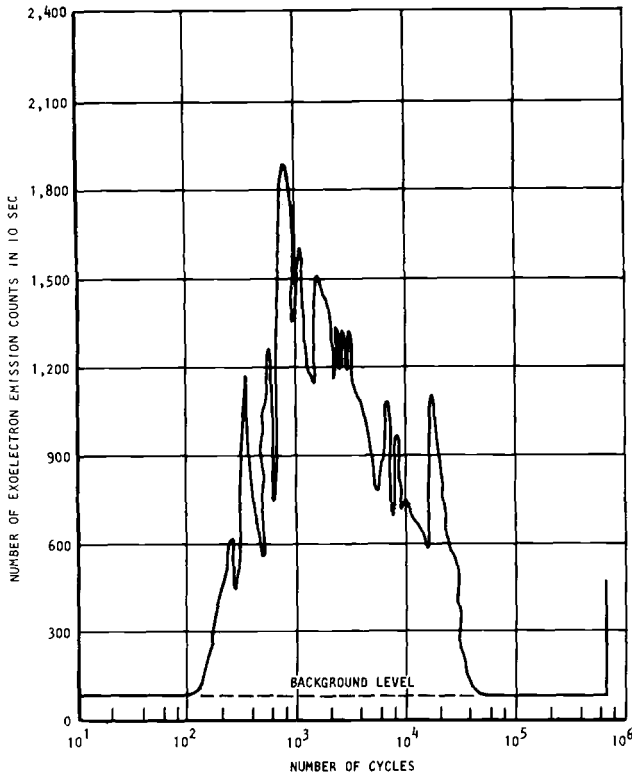


FIG. 3—Exoelectron emission of 1100-O aluminum in fatigue test in vacuum at 11 450 lb/in.².

complete darkness. The recorded intensity did not rise above the initial level, indicating no measurable exoelectron emission from the fatigued specimen. To investigate further the effect of light stimulation, another specimen subjected to stress at 11 450 psi was irradiated alternately in a 15-min period with daylight and longwave ultraviolet light. The emission curve in Fig. 4 shows a considerable increase of intensity during ultraviolet illumination. The vacuum experiments were discontinued in order to investigate the feasibility of performing exoelectron measurements in air.

Three fatigue stress levels (11 400, 11 000, and 10 540 psi) were applied to 1100-O aluminum specimens in air, and the emission current was measured during the fatigue process. Figure 5 shows some typical exoelectron emission curves for specimens at a stress of 11 000 psi. The emission current increased rapidly soon after start of the test, reaching the peak at between 40 and 100 cycles. Thereafter, the emission intensity decreased with the increasing number of cycles until fatigue cracks formed. Then the current increased slowly at first and rose rapidly as the cracks propagated through the test section. The overall shape of emission curve for measurement in air resembles that of the curve for measure-

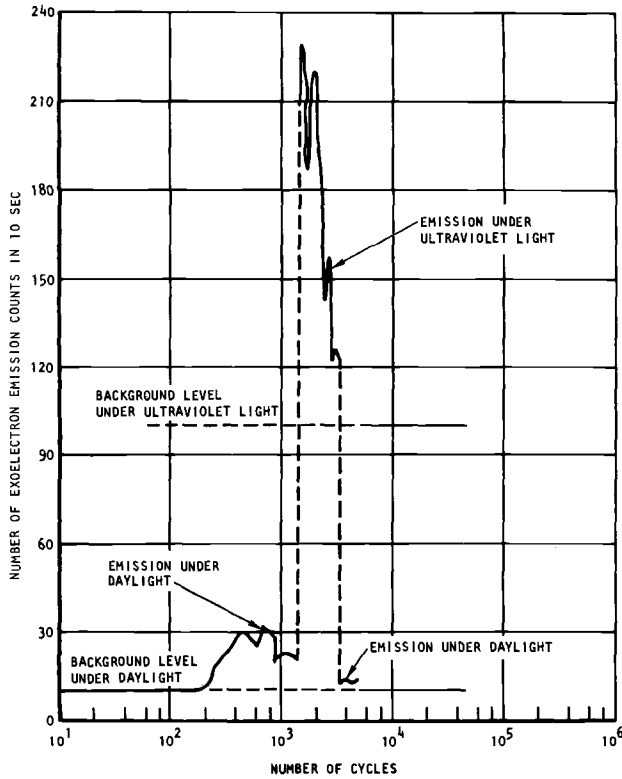


FIG. 4—Exoelectron emission of 1100-0 aluminum in fatigue test in vacuum at 11 450 lb/in.² and under various light conditions.

ment in vacuum, except that the air-measurement curve has only one peak. At stress levels of 10 540 psi and 11 400 psi, the emission curves have characteristics similar to those for the 11 000 psi data. There appears to be no apparent relationship between the strongest emission current during fatigue and the stress applied in the test.

Acoustic Emission

Test Method—The acoustic emission measurements were conducted concurrently with exoelectron emission tests. The acoustic emission transducer was coupled to one face of the test section of the specimen; the other face was illuminated with ultraviolet light to stimulate exoelectron emission.

During the first 1000 cycles, the acoustic emission was recorded in 10- or 100-cycle increments. After 1000 cycles, the emission was recorded at 1000-cycle increments. The acoustic emission for a given fatigue condition was the total number of counts measured in the associated 1000-cycle period.

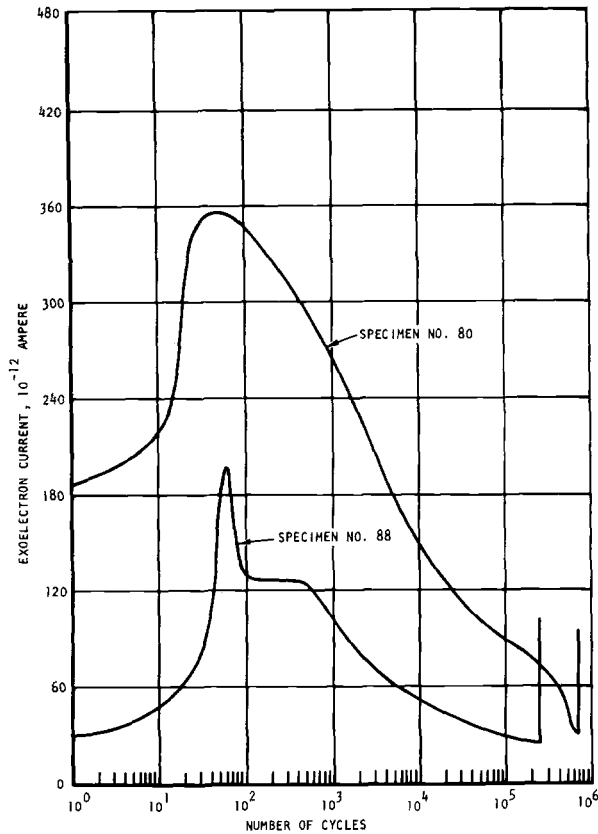


FIG. 5—Exoelectron emission of 1100-0 aluminum in fatigue test in air at 11 000 lb/in.².

Results—During all the fatigue tests on aluminum specimens in air, the signal level of the acoustic emission was very high in the first 100 cycles. This was probably caused by changes in specimen gripping due to the downward movement of the lower grip to accommodate the elongation of the specimen under loading. The emission then gradually decreased and stabilized at approximately 500 cycles.

The acoustic emission intensity, expressed in number of counts in 1000 cycles, could change one order of magnitude several times in the specimen life. These changes also occurred at different stages (in terms of the number of cycles or percentage of total life) in the fatigue process of specimens tested at the same stress level. The initial test results were inconclusive. Subsequently, the test system was modified to reduce electrical noise and mechanical surges in the load cycling system. The acoustic emission measurements were repeated, and the results proved to be more consistent. Two typical curves relating the cumulative number of counts to the number of cycles are illustrated in Fig. 6.

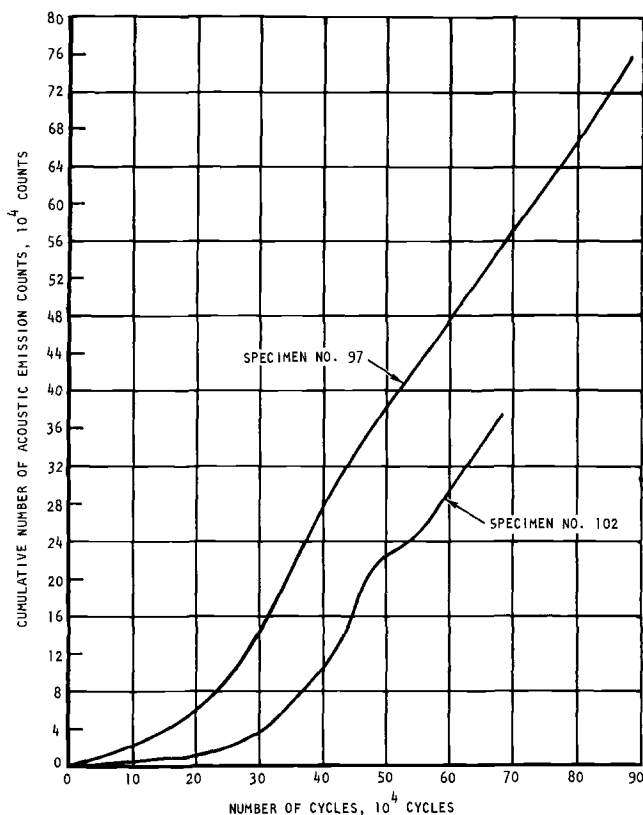


FIG. 6—Acoustic emission of 1100-0 aluminum in fatigue test in air at 11 450 lb/in.².

Metallography

The microstructural change by fatigue process was studied by conducting metallographic examination on electropolished fatigued specimens which had undergone 800, 3×10^4 , and 3.6×10^5 cycles at 11 450 psi. The selected numbers of cycles correspond approximately to the times at which significant changes of exoelectron emission in vacuum took place in the fatigue process. Slip lines or slipbands were evident at as early as 800 cycles. The bands became more numerous as the number of cycles was increased. To gain some idea of the depth affected by the surface slip movement, test sections of five more specimens which had been fatigue stressed at 11 000 psi for 5×10^3 , 1×10^4 , 5×10^4 , 1×10^5 , and 2×10^5 cycles were electropolished in a methanol/nitric acid bath. The polished piece was withdrawn periodically from the bath and examined under an optical microscope. The fatigue affected layer was thought to have been removed by electropolish when no slipbands were evident under the microscope at a magnification of 250, and the microstructure

on the surface reverted to the structure prior to fatigue test. The depth of the removed layer varied between 40 and 75 μm without apparent relation to the number of cycles.

Discussion

Exoelectron Emission

There are several similarities of exoelectron emission behavior of fatigued 1100-0 aluminum in vacuum and in air, among which are: (1) both emission curves exhibit an early rapid rise followed by a rapid, then slower, fall; and (2) there is a final rise in intensity before specimen failure. The fall of emission intensity, however, does not go below the initial level in the vacuum test. This might be attributed to the limitation of sensitivity of the particular electron multiplier.

The initial stage of emission in vacuum has been tentatively defined as a period which ends as soon as the intensity of emission falls to the initial level. Since the presence of this stage is thought to be associated with the limited sensitivity of the measuring instruments, it can hardly be related to the total fatigue life. Furthermore, duration of this stage is very short, being less than 10 percent and one percent of the life at high stress (above 11 000 psi) and low stress (below 11 000 psi), respectively. Changes of emission intensity within this stage cannot be used to predict the happenings in the long remaining safe life.

No such initial emission stage is present in an exoelectron emission curve in air. Once the intensity falls to the initial level, it decreases further with increasing number of cycles before it rises again toward the end of the test. The emission is not apparently influenced by the applied stress. Magnitudes of the emission current can vary appreciably for the specimens subject to equal stressing and the same surface preparation. Regardless of the stress level, the current rises rapidly in the first few hundred fatigue cycles. The metallographic examination of specimens having undergone similar fatigue cycling shows that the aluminum had been only slightly deformed by fatigue, since all traces of slip lines could easily be removed by light etching. Therefore, the rising portion of the emission curve was not included in the analysis. An examination of Fig. 5 shows that, after the elapse of the first 1000 cycles, the emission curves have somewhat similar decreasing characteristics. Therefore, the change of emission current is computed with reference to the current at 1000 cycles, rather than the initial level prior to test.

At a stress level of 11 000 psi, the average fatigue life of the specimen is 3×10^5 cycles. The actual life used in analysis is 2.99×10^5 cycles so that, by the subtraction of 1000 cycles, the reference current is now the value at these cycles. The emission current at various numbers of cycles corresponding to certain percentages of life for the aluminum specimens is noted on the record tape. The percent of current change, with respect

to the value at 1000 cycles, is next computed and plotted versus the percentage of fatigue life in Fig. 7. The same analysis is also applied to the emission curves for stresses of 11 400 and 10 540 psi, and the results are included in Fig. 7.

As mentioned earlier, there is no apparent relation of the strongest emission current with the applied stress, when the current is measured in air. The computed relationship between the change of emission current during fatigue and the spent life also indicates a stress-independent trend. Using this trend, it is possible to predict the remaining safe fatigue life of 1100-0 aluminum by estimating the change of exoelectron emission current in air at the end of the spent life, if the change is based upon the current at the first few hundreds or thousands of cycles when the current decreases steadily after its initial rapid rise and fall.

It should be noted, however, that the measurement of emission current in air must be made while the fatigued specimen is irradiated with ultra-

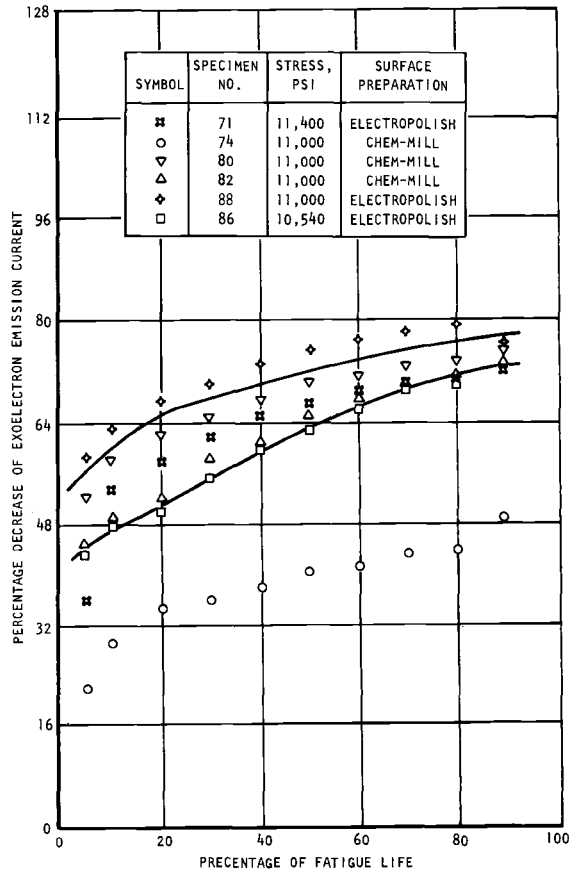


FIG. 7—Relation of percentage decrease of exoelectron current to percentage of fatigue life for 1100-0 aluminum.

violet light. Tests in vacuum indicate that the emission depends very much on light condition: no emission in complete darkness, and considerable increase of emission under ultraviolet light. Evidently, additional energy is required for the electrons to emit from a deformed aluminum surface, as suggested by Pimbley and Francis[23]. Such a relationship between exoelectron emission and incident light energies has also been noticed in abraded and deformed metal surfaces, such as Al, Cu, Ni, Sn, and Zn[24].

Acoustic Emission

The initial results of acoustic emission measurements were inconclusive apparently due to instabilities in the test system. The test system (Fig. 6) was modified and the test results showed a marked change in the slope of the acoustic emission intensity versus fatigue time curve. The change occurs during the early 50 percent of the fatigue life. The significance of this observation is not yet known. Additional measurements are needed at several stress levels to definitely establish a relationship with early fatigue damage.

Fatigue Damage

It was assumed for these experiments that the extent of fatigue damage could be correlated to a percent of fatigue life. The test data for the exoelectron emission and acoustic emission methods appear to relate to the extent of fatigue damage. The stress-independent relationship, shown in Fig. 7 for 1100-0 aluminum, between the change of exoelectron emission current and percentage of fatigue life can be used to estimate the remaining life if the total life is known. The marked change of the slope of the acoustic emission curve, illustrated in Fig. 6 for 1100-0 aluminum, may also be an indicator of early fatigue damage.

The depth of the fatigue affected surface layer can be determined by electropolishing. The depth of fatigue damage for 1100-0 aluminum at a stress of 11 000 psi is about 60 μm .

Summary and Conclusions

Exoelectron emission from 1100-0 aluminum in fatigue was measured in vacuum and in air at room temperature and at stress levels varying between 10 000 and 11 540 psi. The emission was characterized by an early rapid rise and fall in emission intensity followed by a final rise near the end of the test. The final rise is attributed to the formation and propagation of fatigue cracks. The emission intensity in vacuum appears to depend upon the applied stress and is strongly influenced by the incident light energies. For example, no emission was detected in complete darkness. There appears to be a stress-independent relationship between the change of emission current in air, relative to the current after the elapse of a few hundred or a few thousand cycles, and time in fatigue. Illumina-

tion with ultraviolet light is necessary for measuring the emission current in air.

The acoustic emission of 1100-0 aluminum was measured concurrently with exoelectron emission tests in air. The intensity of the acoustic emission undergoes a marked change before 50 percent of the fatigue life has been spent.

Slipbands appear on the specimen surface early in the fatigue life. They do not appear to penetrate deeper as fatigue test proceeds. This indicates that the fatigue damage takes place only at the surface. The fatigue depth of 1100-0 aluminum is probably less than 60 μm , and reaches the limiting value before half the fatigue life at a stress of 11 000 psi has been spent.

Acknowledgments

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Verification of Structural Integrity of Pressure Vessels by Acoustic Emission and Periodic Proof Testing

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ABSTRACT: The application of acoustic emission monitoring of periodic proof tests to the detection of structural degradation of pressure vessels is demonstrated. Test results of welded steel and aluminum vessels are presented which show that structural degradation and impending failure due to fatigue crack growth can be detected by this technique. Subcritical crack growth due to stress corrosion or hydrogen embrittlement could be detected with equal ease. Hence, emission monitoring of a periodic proof test can provide a powerful tool for the verification of the structural integrity of pressure vessels, as well as a wide variety of other structures.

KEY WORDS: predictions, acoustics, emission, pressure vessels, structural analysis, crack propagation, overpressure, degradation, fatigue (materials), stress corrosion, hydrogen embrittlement, plastic deformation

The requirements for safe, long term performance of modern structures and pressure vessels has resulted in the increased need for reliable techniques for the assessment of structural integrity. A technique that utilizes a periodic overpressurization (proof) to levels higher than encountered in service while simultaneously monitoring for acoustic emission, is shown to be applicable for assuring reliability of pressure vessels. This technique is capable of detecting structural degradation in a wide variety of structures, but only the case of pressure vessels will be considered here.

If a pressure vessel has withstood a certain pressure level, it can be expected to operate indefinitely at that level unless some form of structural degradation occurs. This structural degradation is usually in the form of subcritical crack growth, such as fatigue, stress corrosion cracking, or hydrogen embrittlement. Structural degradation can also occur as

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¹ Director of research and president, respectively, Dunegan Research Corp., Livermore, Calif. 94550.

a result of irradiation damage, as well as other conceivable means, but degradation due only to subcritical crack growth will be considered here. Hence, if a pressure vessel has been proofed to a given pressure, and structural degradation occurs only by subcritical crack growth, then the vessel can be considered safe for indefinite use at that pressure if the absence of subcritical crack growth can be guaranteed. Of course, the use of this philosophy is highly dependent on the ability to detect subcritical growth. If this growth can be detected, then the structural integrity of the part can be assessed.

Acoustic emission is a relatively new technique that has been demonstrated to be highly sensitive to the presence and extension of cracks, and is therefore well suited to the verification of the structural integrity of pressure vessels. Since acoustic emission is a relatively new and unknown technique, the following brief discussion will be included. Acoustic emissions are the low level stress waves impulsively generated within the body of a material subject to stress. In metals, it is closely associated with the dislocation movement accompanying plastic deformation, and the initiation and extension of cracks. The detection of acoustic emission is usually accomplished by directly coupling a piezoelectric transducer to the part under test. The output of the piezoelectric sensor is amplified through a high gain, low noise preamplifier, filtered to remove any extraneous low frequency noise, conditioned (in a wide variety of ways) and displayed. Details of the instrumentation used in the present work will be presented in the Experimental Procedure section. An extensive literature on acoustic emission is now available[1-9] which contains details of testing procedures and results. This list of references is by no means exhaustive.

The application of acoustic emission to the detection of structural degradation due to subcritical flaw growth can be accomplished in two ways. The first, and most obvious, is to continuously monitor the part in service to detect the acoustic emissions generated as a result of crack extension. This application of acoustic emission has been demonstrated on laboratory specimens for the detection of crack extension due to fatigue cracking[10,11], stress corrosion cracking[4,12], and hydrogen embrittlement[12,13]. All these studies pointed out the extreme sensitivity of acoustic emission to the presence of subcritical crack growth. However, there are serious drawbacks to continuous monitoring of acoustic emission from in-service pressure vessels and structures. These drawbacks include the excessive background noise that often occurs during service, and the expense involved. Nakamura[14] discusses a technique which can be used to circumvent the excessive background noise problem, and demonstrates its applicability to the continuous monitoring of fatigue crack growth in a noisy environment. This technique shows great potential, but its high cost and complexity could preclude its use in many applications.

The second way that acoustic emission can be used to detect the structural degradation due to subcritical flaw growth is to periodically proof (overpressurize) the vessel while simultaneously monitoring for acoustic emission.² Since acoustic emission is closely tied in with plastic deformation, acoustic emission is irreversible in the same manner as plastic deformation. If a cracked or flawed pressure vessel is pressurized to a given value (say p_p) and then unloaded, no emission will occur during subsequent pressurization until the previous maximum pressure (p_p) is exceeded, unless some form of crack extension has occurred in the meantime. If no crack extension has occurred since the previous pressurization to p_p , and this pressure has not been exceeded in service, then no new plastic deformation, and hence no acoustic emission, will be observed during pressurization to p_p . However, if crack extension has occurred at the working pressure, p_w ($< p_p$) then new plastic deformation, and hence acoustic emission, will be observed while pressurizing to p_p . Furthermore, the more crack extension that occurred during service, the more emission will be observed during the overpressure to p_p . Thus, it is seen that a periodic overpressurization to a proof pressure p_p , with simultaneous acoustic emission monitoring of only the proof test, can be used to detect structural degradation due to subcritical flaw growth. This periodic proof test eliminates the need for continuous monitoring, thereby reducing the expenses involved, and can be performed during periods of low background noise, such as when a vessel is temporarily out of service. Hence, the problem of excessive background noise is circumvented.

The application of this technique to the detection of fatigue crack growth has been demonstrated in laboratory specimens[15,16]. The relatively simple geometries used in these tests allowed a theoretical relationship to be developed relating the acoustic emission observed during the periodic overload to the number of cycles at the working load. Fairly good agreement between the theoretical and experimental results was obtained which shows that quantitative acoustic emission results are attainable. Some preliminary acoustic emission test results of fatigue cycling and periodic overload have been presented by Liptai for tension specimens and pressure vessels of glass-epoxy[17]. Although acoustic emission is capable of detecting damage in composite materials, the test results on these materials do not provide the early and clear warning of impending failure that has been observed on metal specimens.

The purpose of the present paper is to demonstrate the applicability

² The discussion in this paper assumes that the emission is irreversible in the time scale of interest in the test series. Hence, time and aging effects (such as annealing of a vessel operating at elevated temperatures) are ignored. The validity of this assumption should, of course, be evaluated before attempting to apply acoustic emission in the manner described in this paper to a part in which the omission could possibly be reversible.

of acoustic emission monitoring and a periodic overload to the detection of structural degradation due to subcritical crack growth in a situation more related to practical service conditions than the laboratory studies mentioned above. Cyclically pressurized welded metal pressure vessels were selected for this investigation. This technique is applicable to the detection of crack growth due to mechanisms other than fatigue, to structures other than pressure vessels, and to materials other than metals, but this paper will be restricted to these particular cases.

Experimental Procedure

A block diagram of the experimental set up used in this investigation is presented in Fig. 1. A counter was used to provide the means of measuring the acoustic emission activity. The counter counts the number of

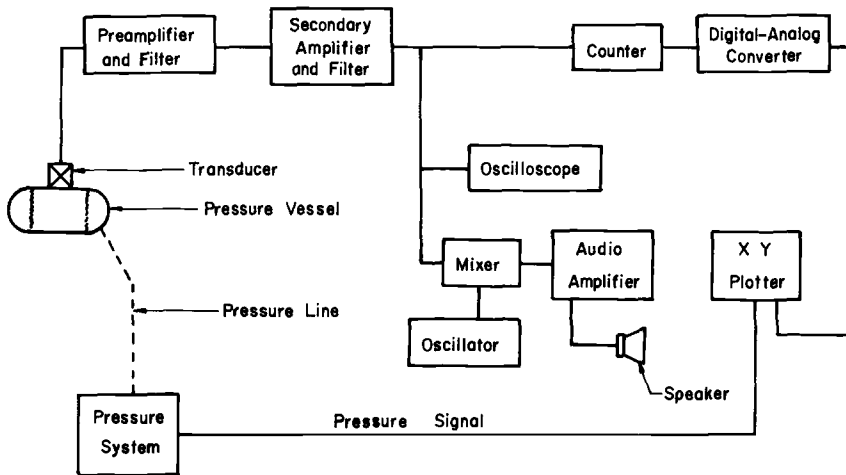


FIG. 1—Block diagram of acoustic emission instrumentation and pressurization system.

times the amplified and filtered output of the sensor exceeds a certain trigger voltage. A digital-analog converter was used to provide a d-c voltage signal proportional to the number of counts. This d-c voltage was inserted into an x-y plotter so that the emission as a function of pressure could be plotted as the test was in progress. A "differential" transducer was used in all the tests, along with a gain of 85 dB and bandpass of 70 to 170 kHz. The pressure system was "closed loop" with provisions for fatigue cycling or ramp pressurization.

With minor variations a single test series consisted of the following steps:

1. Initial pressurization, with acoustic emission monitoring, to a pressure level designated as the proof pressure, p_p . The pressure was held

at this level until the emission died out, at which time the pressure was released.

2. A second pressurization with emission monitoring to this same level, with a hold, and subsequent release of the pressure.

3. The vessel was then cycled at a working pressure, p_w , for a given number of cycles. Acoustic emission activity was not monitored during this period. The working pressure was taken to be $\frac{2}{3}$ the proof pressure ($p_w = 0.67 p_p$). A cycling rate of roughly 1 Hz was utilized.

4. The vessel was then proofed to p_p with simultaneous acoustic monitoring. Emission as a function of pressure was recorded, as well as the number of counts observed while holding at p_p . The vessel was held at the proof pressure until the emission died out.

5. The cycling with intermittent proofing (steps 3 and 4) was continued until failure of the vessel.

Two series of tests were conducted, both on welded cylindrical pressure vessels. They were fabricated by welding end caps onto a 6-in. length of 4-in. OD, $\frac{1}{4}$ -in. wall pipe. One of the vessels was 6061 aluminum which was heat treated to the T6 condition after welding. The weld was purposefully made with only partial penetration. The other vessel was 4130 steel, which was quenched and tempered to a Rockwell "C" hardness of 34 after the end caps were welded on. A circular arc flaw $\frac{1}{8}$ in. deep and $1\frac{1}{4}$ in. long was machined in the central portion of this vessel with a milling cutter. The flaw was oriented in the axial direction; hence, the hoop stress tended to open the flaw.

Experimental Results and Discussion

Aluminum Vessel

The results of the initial pressurization and immediate repressurization of the aluminum vessel are presented in Fig. 2. The emission activity observed up to the 2000 psi level indicated that this pressure was an appreciable fraction of the failure pressure of the vessel, and that the vessel would fail in a convenient number of cycles. Hence, 2000 psi was selected as the proof pressure. The amount of emission during the first repressurization was very small—as expected due to the irreversibility of the emission. Some activity was observed while holding at the proof pressure during the repressurization, which provided further evidence that this pressure was an appreciable fraction of the failure pressure of the vessel.

This vessel was cycled at 1300 psi, and proofed to 2000 psi every 3000 cycles (or 6000 cycles after 30 000 cycles were achieved). The emission activity as a function of pressure for these periodic proof tests is presented in Fig. 3, along with the amount of emission observed while holding at 2000 psi. The numbers by the test results are the number of

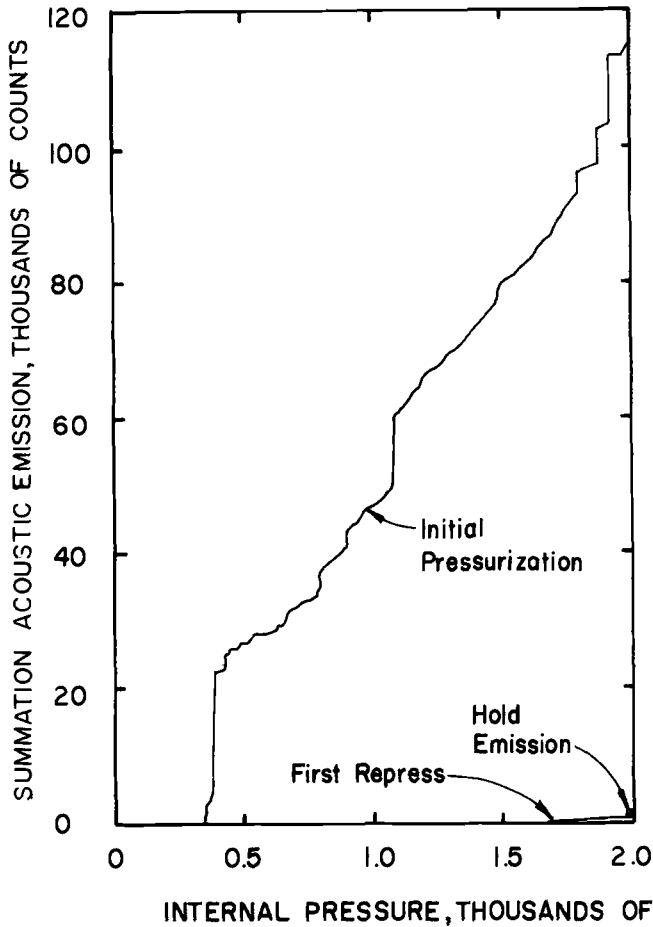


FIG. 2—Summation acoustic emission as a function of pressure for initial pressurization and first repressurization of aluminum pressure vessel.

fatigue cycles in thousands of cycles. These test results show that consistent and extensive fatigue crack growth had begun prior to 42 000 cycles. This is evidenced by the consistent build up in both the emission activity during the rising load portion of the test, and the amount of emission observed while holding at the maximum pressure. This build up is shown more clearly in the data presented in Table 1, which includes the total amount of emission observed during the rising load portion of the test, as well as the hold emission. The data of Table 1 are plotted in Fig. 4, which also clearly shows the consistent build up in emission that occurred after the proof at 36 000 cycles. The vessel failed catastrophically in one of the welds during the rising load portion of the proof test conducted after 51 000 cycles. Thus, it is seen that this technique is capable of detecting subcritical flaw growth, and is capable of providing early

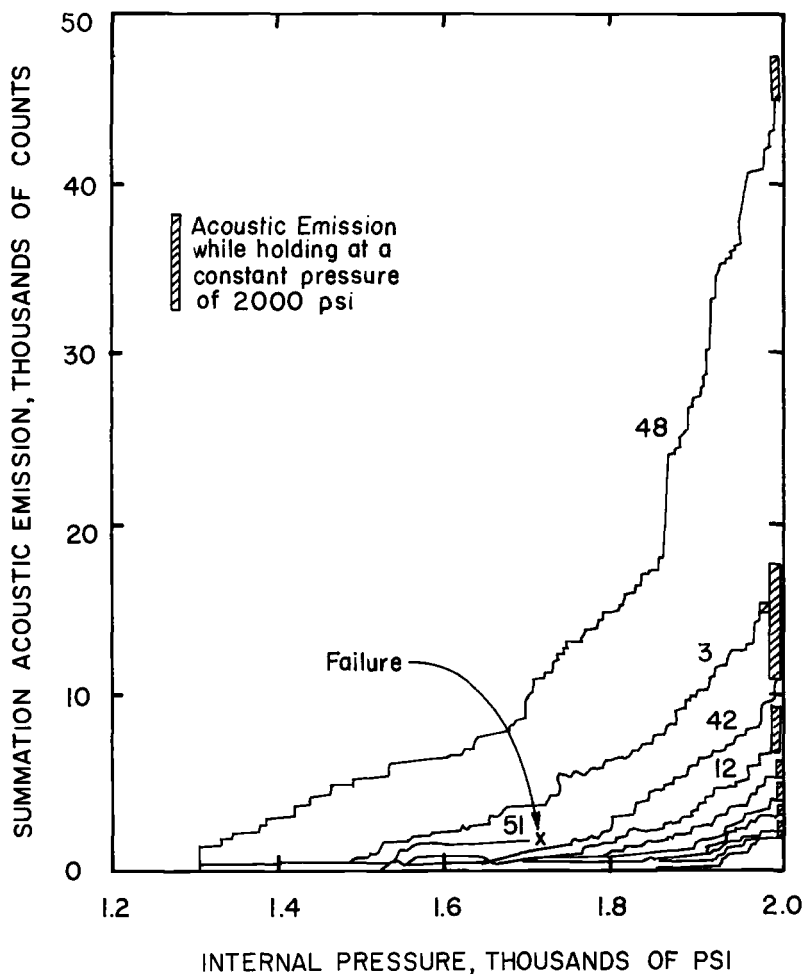


FIG. 3—Summation acoustic emission as a function of pressure for intermittently proofed cyclically loaded aluminum pressure vessel. Numbers by results are the number of fatigue cycles in thousands of cycles.

and ample evidence of impending failure; the impending failure was detected at 80 percent of the life of the part. Figure 5 presents photographs of the failed vessel, showing both an overall view of the vessel, and a close up of the fracture surface showing the fatigue markings due to the periodic overload.

Steel Vessel

The results of the initial pressurizations for this vessel are presented in Fig. 6. This vessel was initially pressurized to 12 000 psi. The emission activity indicated that this would be a convenient proof pressure. Relatively little emission was observed during the first repressurization (as

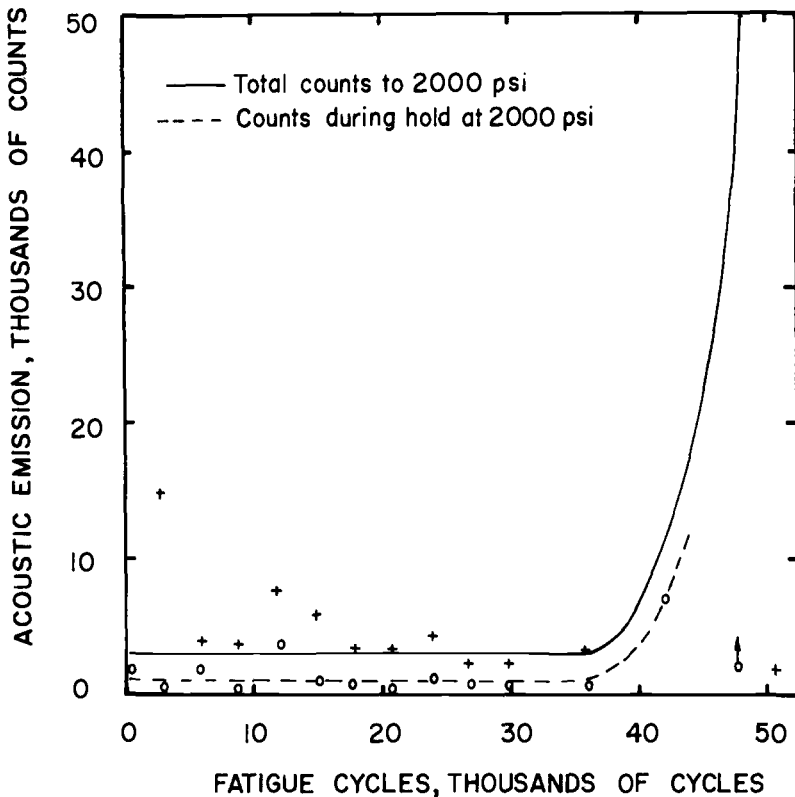


FIG. 4—Total acoustic emission observed during proof to 2000 psi as a function of the number of fatigue cycles to 1300 psi showing the build up in activity prior to final failure during the proof after 51 000 cycles.

TABLE 1—Experimental results for aluminum vessel.

Number of Fatigue Cycles, Thousands of Cycles	Total Counts to 2000 psi, Counts	Counts During Hold At 2000 psi, Counts
0	200	2 000
3	14 945	390
6	3 890	1 910
9	3 570	644
12	7 532	3 591
15	6 018	1 156
18	3 469	984
21	3 354	489
24	4 209	1 181
27	2 435	975
30	2 412	750
36	3 156	890
42	11 243	7 040
48	45 210	2 312 ^a
51	1 810 ^b	...

^a Hold test terminated early, so this result is not meaningful. Much emission during hold.

^b Failed during proof loading at 1715 psi.

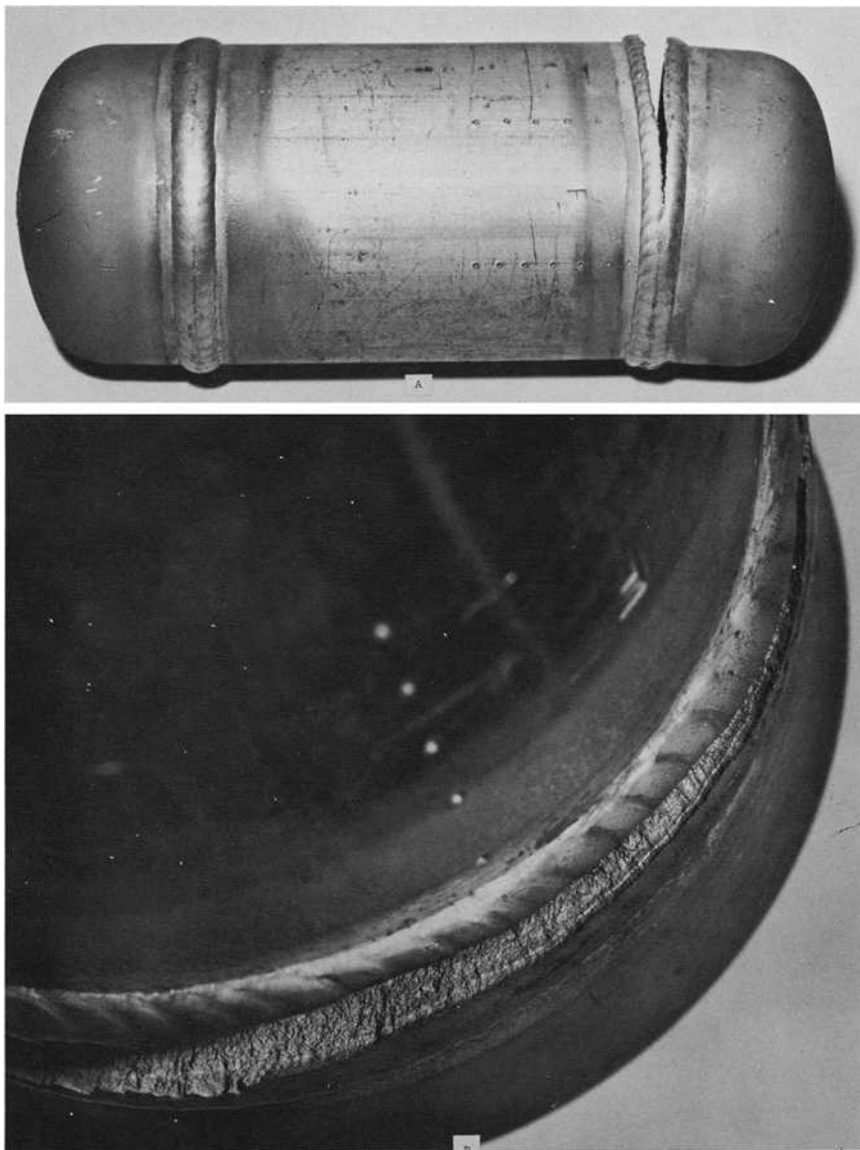


FIG. 5—*Photographs of failed aluminum vessel: (a) overall view, (b) fracture surface showing fatigue markings due to the periodic overload.*

expected). The vessel was then cycled 5000 times to 8000 psi, and then proofed. As shown in Fig. 6, the emission activity increased very rapidly as the pressure exceeded the cyclic value of 8000 psi. This very rapid build up is clear evidence of impending failure. It is felt that the vessel would have failed prior to reaching the proof pressure of 12 000 psi. The

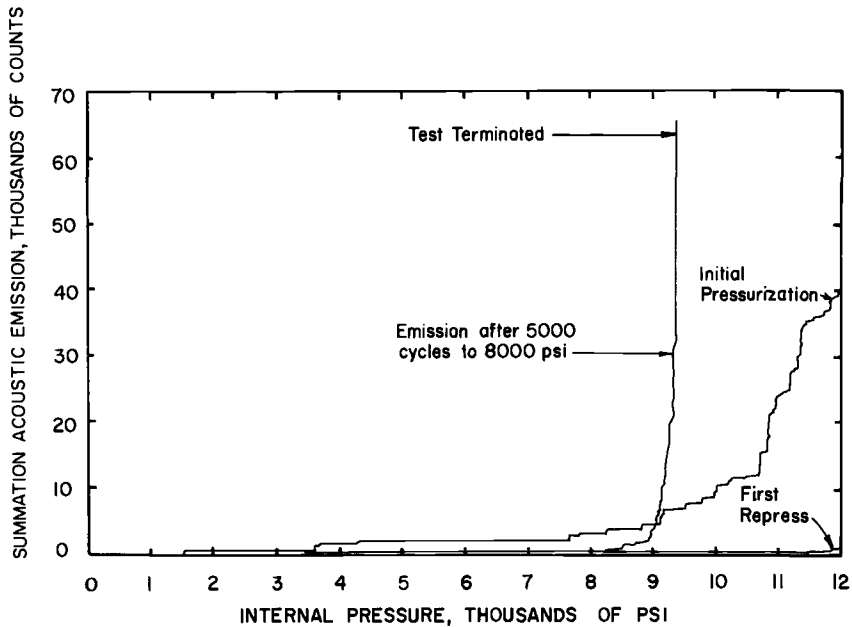


FIG. 6—Summation acoustic emission as a function of pressure for initial pressurization, first repressurization, and first proof test of steel pressure vessel.

pressure was therefore released, and a new set of pressure parameters selected.

The vessel was then cycled 22 000 times to 2000 psi and 45 000 times to 4000 psi with no build up in the emission during the periodic proof test. It was then cycled 5000 times to 5000 psi, and emission was observed during the proof to 7500 psi. Hence, this was taken as the new proof pressure. The results of the series of tests conducted by cycling to 5000 psi with a proof test every 5000 cycles are presented in Fig. 7. This figure shows the increasing amount of emission observed during the proof after 10 000 cycles, which indicated that the crack was now growing.

The proof test after 15 000 cycles to 5000 psi consisted of two parts, as indicated in Fig. 7. The first part was terminated and the pressure released before reaching the 7500 psi proof level due to a malfunction of the pressure system. The results for the first proof at 15 000 cycles provides clear evidence of extensive structural deterioration of the vessel. After repairing the pressure system, the vessel was again proofed. The emission during this second pressurization began at about 6900 psi, which is considerably below the 7400 psi the vessel had just been subjected to. This reversibility of the emission provided strong evidence of impending failure. The test was continued on to the proof pressure of 7500 psi, with a great deal of emission being observed. The pressure was held at this level, and the emission continued to build up. After about 2 min at 7500

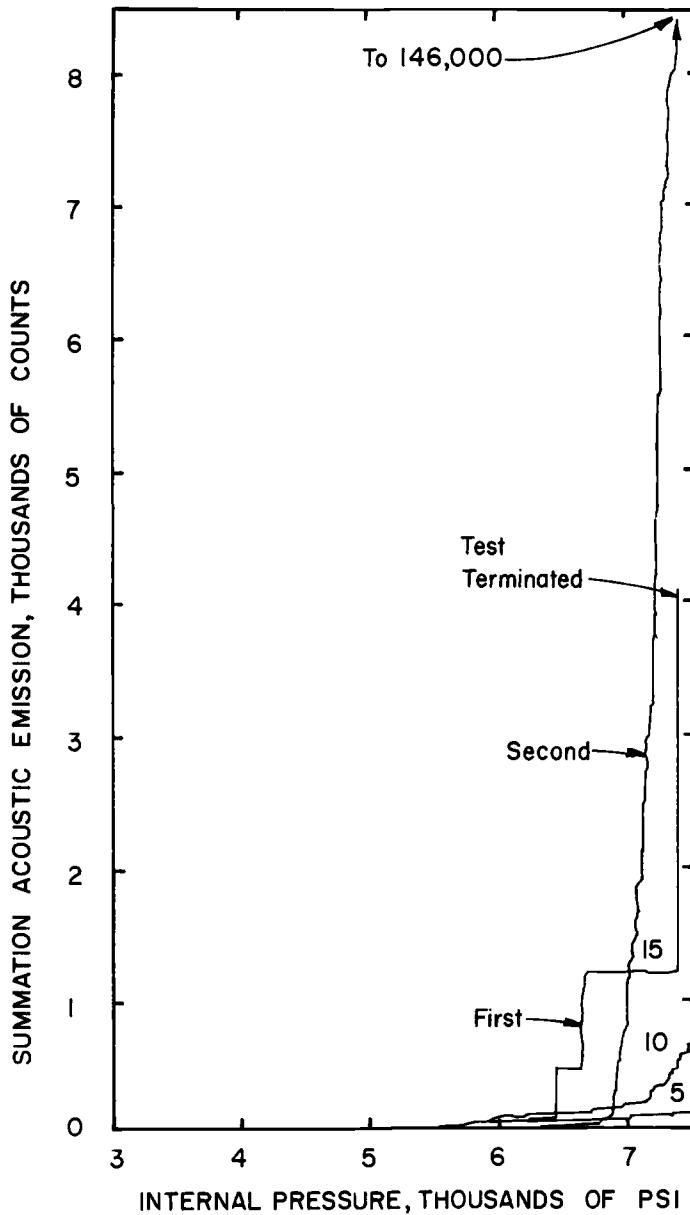


FIG. 7.—Summation acoustic emission as a function of pressure for intermittently proofed cyclically loaded steel pressure vessel. Numbers by results are the number of fatigue cycles to 5000 psi in thousands of cycles. See text for explanation of "first" and "second" proofs after 15 000 cycles.

psi, and a great deal of emission, the vessel started to leak. This constituted failure, and the test was terminated.

These test results indicate that a fatigue crack was growing during the cyclic pressurization to 5000 psi. Once again, clear evidence of impen-

ing failure was obtained thousands of cycles prior to final failure; the impending failure was detected at about 70 percent of the life of the part. The copious emission during the proof tests after 15 000 cycles indicated that failure was imminent. The extensive emission observed during the final proof indicated that the crack was growing under constant load, which was definitely the case, since the vessel sprung a leak at that time.

The flawed portion of the vessel was cut out, immersed in liquid nitrogen, and broken. Figure 8 shows the resulting fracture surface, which clearly delineates the crack growth due to the fatigue loading that eventually resulted in the leak.

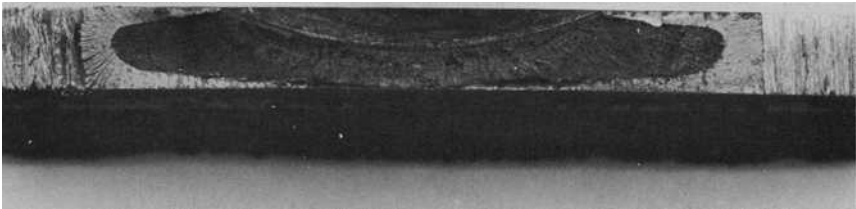


FIG. 8—*Photograph of the surface of the through-the-thickness flaw that grew from the machined notch and resulted in leak.*

Conclusions

The test results presented show that acoustic emission monitoring of a periodic proof test is capable of providing early and ample warning of impending failure due to subcritical crack growth. The impending failure was detected at 70 to 80 percent of the life of the part in the two vessels studied. Hence, this technique is well suited for the verification of structural integrity of pressure vessels, if structural degradation occurs as a result of subcritical growth. The two series of tests presented were for widely varying conditions. Aluminum and steel vessels with crack growth in both weld and base material were studied and final failure was by leaking in one, and catastrophic crack growth in the other. In both instances, evidence of structural degradation was obtained in two manners: increasing amounts of emission during the rising load portion of the test, and increasing emission while holding at the proof pressure. Hence, it can be concluded that acoustic emission monitoring of a periodic overload is capable of detecting structural degradation in a wide variety of materials, and can therefore provide a powerful tool for the verification of the integrity of not only pressure vessels but a wide variety of other structures.

Acknowledgments

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Empirical Strength Theories

REFERENCE: Sendeckyj, G. P., "Empirical Strength Theories," *Testing for Prediction of Material Performance in Structures and Components,* ASTM STP 515, American Society for Testing and Materials, 1972, pp. 171-179.

ABSTRACT: The formulation of empirical strength criteria is considered and some general principles that should be satisfied by all strength criteria are discussed. The available strength criteria for isotropic materials, with unequal strengths in tension and compression, are discussed in the light of the general principles. Some recent criteria for anisotropic materials are also presented. Contributions from the Russian literature are extensively referenced.

KEY WORDS: predictions, criteria, strength, isotropy, anisotropy, shear strength, failure, loads (forces), stresses, strains

Considerable effort has been expended recently in determining the strength of composite materials (see for example the extensive bibliographies in Refs 1-5). This effort has been directed in two principal directions, namely, experimental determination of the strength under various loading conditions and theoretical prediction of strength based on postulated failure models. The present work will be restricted to the determination of strength based on experimental results.

Because a purely experimental determination of strength, consisting of performing tests for all possible loading conditions that might be encountered in actual structures, is impossible since one would have to perform a prohibitively large number of tests, the commonly accepted practice is to postulate a strength criterion containing a number of parameters that can be determined by conducting a small number of tests. A portion of the experimental data is then used to fix the parameters, while the remaining test data are used for verification purposes.

The ease of postulating empirical strength criteria and running tests

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¹ Assistant professor of engineering, Division of Solid Mechanics, Structures and Mechanical Design, Case Western Reserve University, Cleveland, Ohio 44106. Presently at Advance Composites Branch, Structures Division, Air Force Flight Dynamics Laboratory, Wright-Patterson Air Force Base, Ohio 45433.

for their verification has led to a proliferation of strength theories and the large selection that is possible from among these criteria has led to considerable confusion. Which theory from among the proposed ones should be used when dealing with a new material? Is it possible to decide the applicability of a given strength criterion without performing at least the minimum number of tests necessary for fixing the parameters of the criterion? Does the researcher have to discover the hidden pitfalls in the various theories for himself?

An attempt to give a unified treatment of strength criteria and to answer the above questions is made in the present work. Some general principles that should be satisfied by empirical strength theories are discussed in the next section. A brief review of various theories that have been proposed for isotropic and anisotropic materials is then given. Finally, the possible pitfalls of poor quality testing are illustrated by an example from the literature.

General Form of Empirical Strength Criteria

In the formulation of empirical strength theories, it would seem at first that any relationship between stresses or strains or both would serve as a possible strength criterion. A criterion, in which both the stresses and strains appear, can be eliminated from consideration due to its relative complexity when compared to one in which only the stresses or strains appear. The selection of stresses as the basic variables in a strength theory is made for convenience and not because the stresses are more appropriate than the strains.

However, a little reflection on possible strength characteristics of a solid leads to the conclusion that not all possible relations between stresses lead to physically meaningful results. For example, it has been observed that most materials exhibit essentially infinite strength when loaded in equal triaxial compression. Thus a strength criterion to be valid must predict the infinite strength under this loading condition. This requirement has the effect of imposing a restriction on the form of the strength criterion, but other restrictions can also be imposed.

Restrictions on Strength Criteria

A number of principles can be used to establish the form of a strength criterion. These can be stated as follows:

1. The strength criterion must be invariant with respect to coordinate transformations. Otherwise, a change of description would produce a different strength in the same material subjected to the same stress state. Coordinate invariance is satisfied by writing the strength criterion in tensor form.

2. The strength criterion must satisfy Drucker's postulate; that is, the strength surface (a plot of the limiting values of strength in a nine-

dimensional stress space) must be convex. Even though this requirement is used in formulating yield criteria, there exist some questions about its application in formulation of failure criteria. In the present work, it will be used to limit the class of possible strength criteria.

3. The criterion must predict the strength peculiarities of the material under consideration. The criterion must be able to predict different strength behavior of the material in tension and compression. It must also be able to predict the dependence of strength on orientation of the test specimen with respect to planes of material symmetry. In particular, the criterion must exhibit the proper behavior for simple test specimen configurations.

4. The criterion should be as simple as possible. A strength criterion to gain wide acceptance must be either easy to use or must be simple in form. Furthermore, the number of experimentally determined parameters needed for characterization of the strength should be kept to a minimum.

Isotropic Materials—Unequal Strength in Tension and Compression

In light of the principles stated above, it is interesting to review some of the previously proposed empirical strength criteria for isotropic materials. The classical theories (maximum normal stress, maximum shearing stress, maximum strain, Coulomb and Mohr criteria, etc.) have been discussed thoroughly by a number of authors[6,7] and, hence, they will not be considered here.

As the desire for more accurate theories developed, the proposed criteria proliferated with the result that one can find dozens of criteria dealing with isotropic materials exhibiting different strengths in tension and compression. Yet most of these fall into two broad categories based on the description of the strength surface in the Haigh-Westergaard stress space (a space in which the principal stresses are taken as coordinate axes). The first group of criteria describe surfaces of revolution about the hydrostatic axis which is defined by the condition $\sigma_1 = \sigma_2 = \sigma_3$ where σ_i ($i = 1, 2, 3$) are the three principal stresses. The possibility of constructing such criteria was apparently first suggested by W. V. Burzynski[8] and A. Nadai[9]. Nadai proposed that the strength criterion can be written in the general form

$$\tau_0 = f(\sigma_0) \quad (1)$$

where τ_0 and σ_0 are respectively the octahedral shearing and normal stresses defined by

$$\tau_0 = [(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2]^{1/2}/3 \quad (2)$$

$$\sigma_0 = (\sigma_1 + \sigma_2 + \sigma_3)/3 \quad (3)$$

It is convenient to introduce the invariants of the stress tensor

$$\begin{aligned} I_1 &= \sigma_1 + \sigma_2 + \sigma_3 \\ I_2 &= \sigma_1\sigma_2 + \sigma_2\sigma_3 + \sigma_3\sigma_1 \\ I_3 &= \sigma_1\sigma_2\sigma_3 \end{aligned} \quad (4)$$

and of the stress deviator tensor

$$\begin{aligned} J_1 &= 0 \\ J_2 &= [(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2]/6 \\ J_3 &= -[3(\sigma_1\sigma_2^2 + \sigma_2\sigma_3^2 + \sigma_3\sigma_1^2 + \sigma_2\sigma_1^2 + \sigma_3\sigma_2^2 + \sigma_1\sigma_3^2) \\ &\quad - 12\sigma_1\sigma_2\sigma_3 - 2(\sigma_1^3 - \sigma_2^3 + \sigma_3^3)]/27 \end{aligned} \quad (5)$$

In terms of the invariants,

$$\sigma_0 = I_1/3 \quad (6)$$

$$\tau_0 = [2I_1^2 - 6I_2]^{1/2}/3 = [2J_2/3]^{1/2} \quad (7)$$

Using this notation, the various strength criteria can be written rather succinctly.

Criterion of Schleicher[10]

In 1925, Schleicher proposed that the strength of an isotropic solid can be predicted by taking the total energy of deformation as a function of the hydrostatic stress. He considered the case of a linear dependence and his criterion can be written as

$$J_2/\sqrt{2} - 2\mu I_2 + (C - T)I_1 \leq CT \quad (8)$$

where C , T , and μ are the uniaxial compressive strength, uniaxial tensile strength and the shear modulus, respectively. The shear strength, S , is given by

$$S = [CT/2(1 + \mu)]^{1/2}$$

This criterion has two main drawbacks, namely, the shear strength depends on the shear modulus of the material and the criterion predicts a finite strength in the case of equal triaxial compressive loading.

Criterion of Burzynski[8] and Special Cases

In 1929, Burzynski proposed that the strength can be described by

$$J_2 = A + BI_1 + DI_1^2 \quad (9)$$

where the constants A , B , D are determined from knowledge of the strength in uniaxial tension, uniaxial compression, and shear. It should be noted that combinations of the constants A , B and D exist for which the criterion violates Drucker's postulate. In fact such a particular case has been suggested as a strength criterion by Mirolyubov[11].

In 1931, Y. N. Yagn[12] arrived at criterion (Eq 9) by starting from a somewhat different hypothesis. P. P. Balandin[13] assumed that the energy of deformation is a function of the hydrostatic stress. In particular, he considered the case of a linear dependence on the hydrostatic stress, which is a special case of Eq 9.

Other Criteria of the Type in Eq 9

The literature abounds with criteria that can be put in the form of Eq 9. Some typical examples are:

$$\sqrt{\sigma} J_2^{1/2} \leq 2\sqrt{2} CT/(C + T) - \sqrt{2}(C - T)I_1/(C + T) \quad (10)$$

due to A. I. Botkin[4]; and

$$\tau_0 = (a - bI_1)^2 \quad (11)$$

due to Drucker and Prager[14]; and

$$\tau_0 = (a - bI_1)^2 - CI_1^2 \quad (12)$$

proposed by Matsuura[15] as a generalization of (11); and

$$\tau_0 = a + bI_1 + CI_1^2 \quad (13)$$

proposed by Hu and Pae[16]; and

$$aI_1 + bI_1^2 + C\tau_0 = 1 \quad (14)$$

given by Kan and Pervushin[17]. It should be noted that Eq 14 can also be written in the form of Eq 13. Finally, Gol'denblat and Kopnov[2] suggested that the strength is governed by

$$(1/T - 1/C)I_1/2 + \{\sigma_{ij}\sigma_{ij}/2S^2 + [(1/T + 1/C)^2/4 - 1/2S^2]I_1^2\}^{1/2} \leq 1 \quad (15)$$

where σ_{ij} is the stress tensor and repeated subscripts denote summation over them. The constants T , C , S in Eq 15 must satisfy the inequalities

$$K = g(1/T + 1/C)^2/4 - 3/S^2 \geq 0 \quad (16)$$

$$K^{1/2} \geq 3|1/T - 1/C|/2$$

otherwise, the criterion does not make sense.

More General Strength Criteria

A number of strength criteria, which cannot be described as surfaces of revolution, have been proposed recently. Among these we have

$$J_2/2 = [TC + (C - T)I_1]\{1 - (1 - 3S^2/CT)[1 - (J_2/18)^{-3/2}I_3/2]\}, \quad (17)$$

proposed by Geniev and Kissyuk[18], for which Drucker's postulate imposes the condition

$$(11/39)^{1/2} \leq S/(TC)^{1/2} \leq 1/\sqrt{3}$$

criteria of Tarasenko[4],

$$\sigma_1 + aI_1 + bI_1 \sin \pi \mu_\sigma = \text{constant} \quad (18)$$

where

$$\mu_\sigma = (2\sigma_2 - \sigma_1 - \sigma_3)/(\sigma_1 - \sigma_2) \quad (19)$$

is the Lode parameter;

$$J_2 = a + b\sigma_0 + d(1 - \cos 2\pi\mu_\sigma) \quad (20)$$

proposed by Yagn and Vinogradov[19]; and

$$J_2^\alpha + \sigma_1^\beta (C^\alpha - T^\alpha)/T^\beta \leq C^\alpha \quad (21)$$

where α and β are integers, proposed by Pisarenko and Lebedev[20]. In particular, they considered the cases: $\alpha = \beta = 1$, $\alpha = 2$ and $\beta = 1$, and $\alpha = \beta = 2$. A geometrical interpretation of this criterion is given in [21]. Extensive testing of the criterion has been carried out[4,22] with the conclusion that it best represents the experimental data.

Finally, it should be mentioned that various generalizations of the Coulomb-Mohr criteria exist[23,24].

Anisotropic Materials

Even though the development of high performance composites has only recently led to considerable interest in strength criteria for anisotropic materials, the literature goes back to the early work in the wood industry[25,26]. Most of the early endeavors deal with the so-called interaction equation, giving the angular dependence of uniaxial strength. Werren[26] showed that the strength can be predicted by setting a limit on the energy of distortion stored in the material.

A similar criterion was proposed by Hill[27] and it was first used for composites by Tsai[28]. The Mises-Hill strength criterion gives good results for fiberglass. A factor detracting from its usefulness is that it assumes that the material has tensile and compressive strengths that are equal for all orientations of the specimen. Furthermore, Ashkenazi[25] has shown that it can lead to contradictory results for highly anisotropic materials. A number of modifications have been proposed which will not be discussed here (for a discussion of them, see Refs 1,2,5,7,9, and 29).

Only the criteria proposed by Gol'denblat and Kopnov[2] and Ashkenazi[1,25] will be mentioned here. For a discussion of the other criteria the reader is referred to[29]. Gol'denblat and Kopnov, in an attempt to provide a systematic development of strength criteria for anisotropic solids assumed that the strength is governed by the following equation:

$$[A_{ij}\sigma_{ij}]^\alpha + [A_{ijmn}\sigma_{ij}\sigma_{mn}]^\beta + \dots \leq 1 \quad (22)$$

where A_{ij} , A_{ijmn} , . . . and α , β , . . . are constants that have to be determined. Imposing the condition that the criterion must be homogeneous in the stresses, they got $\alpha = 1$, $\beta = 1/2$, ... For the sake of simplicity they took only the first two terms in Eq 22. The number of independent constants A_{ij} and A_{ijmn} were restricted by assuming that the material possesses the same symmetry properties for strength as it does for the elastic constants. Working equations are given in[2].

As an alternate approach, Ashkenazi[1,25] built up a strength crite-

tion from a number of simple considerations. First, by assuming that the angular dependence of strength in uniaxial tension and compression possesses the same orientation dependence as the elastic constants, Ashkenazi obtained criteria valid for a simple uniaxial stress state. Next, a quartic equation, containing extra parameters, was constructed such that for uniaxial loading it reduces to the correct results. Finally, the parameters were fixed by running tests under combined loading. (This criterion is rather difficult to use because of the large number of tests that are needed to fix all the parameters.)

Experimental Determination of Parameters

The acceptance of a given strength criterion depends on a number of factors, namely, the simplicity of the criterion, simplicity and ease of performing tests to determine the parameters of the criterion, and the quality of fit of the experimental data. It is obvious that a simple criterion containing two parameters is preferable to one containing three, provided that the fit of the experimental data is equally good. Thus, when dealing with a new material, the experimenter can be expected to try strength criteria in the order of increasing number of parameters. If a criterion with a small number of parameters seems to fit the experimental data well, the tendency is to accept it and to discontinue testing. This seems to be particularly true in the area of composite materials where the initial batches of a new composite are extremely expensive and the tester is the developer of the material. This underlies the main difficulty faced by the user of such materials.

A user, who is unable to carry out a complete test program on a new material, is faced with the problem of interpreting the data obtained by others. His job is complicated by the proliferation of experimental papers, many of which are of dubious quality, since the data are incomplete. A prime example of this type of paper is any paper dealing with composites that does not report the void content of the material, number of tests run, or details of the testing procedure. In fact, there exists a lot of controversy about the validity of experimental data and test specimen configuration. This difficulty can be resolved by standardizing test specimens provided that reliable analysis exists for the specimen. An effort in this direction has been undertaken by a number of authors (see for example Refs 30-34).

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Predicting the Service Life of Neoprene Launch Tube Liner Pads for the Poseidon Missile

REFERENCE: Rudd, G. E., Meier, J. F., and Rosenblatt, G. B., "Predicting the Service Life of Neoprene Launch Tube Liner Pads for the Poseidon Missile," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 180-197.

ABSTRACT: Molded neoprene liner pads used to support the Poseidon missile in the launch tube were found to stiffen with time. Increasing the postcure reduced the stiffening rate. By aging pads at several moderately elevated temperatures it is possible to predict the service life of pads with several different postcures. The results of such aging tests, including the predicted service lives, are presented in this paper.

KEY WORDS: predictions, chloroprene resins, age hardening, stiffening, missiles, linings, padding, crosslinking, oxidation, tests, carbon black, strains

Molded neoprene liner pads support the Poseidon missile in the launch tube lower region as shown in Fig. 1. The pads provide shock mitigation, react with the missile during launch, and damp vibrations. The upper region pads are of cast polyurethane and are in contact with the missile during loading and ejection only. The concept, design, and materials for both types of pads were developed by Westinghouse[1-3].

Early neoprene pads were found to stiffen with age and a program was undertaken to minimize the stiffening and to predict the stiffening rate of existing pads. It was found that the stiffening rate decreased with increased pad postcure. It was already known that absolute stiffness increased with postcure and decreased when the amount of FT carbon black in the formulation was reduced. The latter relationship makes it possible to incorporate a postcure in a production process and compensate for the resulting increased stiffness by reducing the carbon black

¹ Senior engineer and senior scientist, respectively, Westinghouse Electric Corp., Research Laboratories, Pittsburgh, Pa. 15235.

² Advisory engineer, Westinghouse Electric Corp. Missile Launching and Handling Dept. Sunnyvale, Calif. 94088.

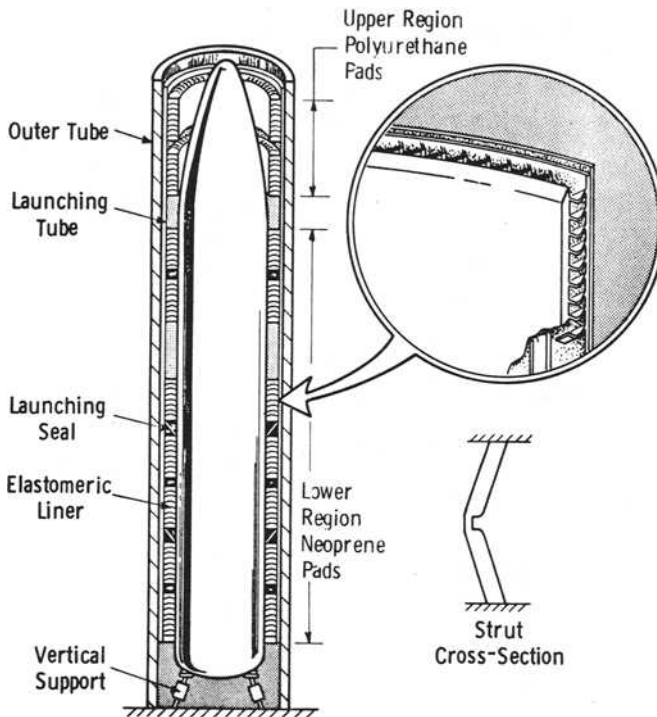


FIG. 1—Schematic launcher tube.

content. The relationship between postcure and carbon black content, as regards absolute stiffness, is illustrated in Fig. 2 where static compression-deflection (C-D) curves for three carbon black loadings and three postcure times are given. The complete formulation and hardness of the rubber are given in Table 1.

TABLE 1—Neoprene rubber pad formulation.

Neoprene WRT	100 phr ^a	An elastomer composed of polymerized 2 chloro-1, 3 butadiene
Maglite D	5.3	Magnesium Oxide, Activator
Neozone A	2.8	Phenyl-alpha-naphthylamine, Antioxidant
Wingstay 100	2.0	Mixed Diaryl Paraphenylenediamines, Antioxidant, Antiozonant
Stearic Acid	0.5	Dispersing Agent
FT Carbon Black	140 (variable)	Reinforcement and Stiffening of Compound
Sundex 790	2.6	Oil, Processing Aid
Rubberol	2.6	Rubber Lubricant and Plasticizer, Aids in Demolding
Zinc Oxide	5.3	Scorch and Cure Rate Control
Monex	0.66	Tetramethylthiuram monosulfide, Accelerator
DOTG	0.5	Diorthotolylguanidine, Curative
Sulfur	1.0	Curative
Hardness, Durometer A ^b	81-86	

a phr is the abbreviation for parts per hundred parts rubber by weight

b ASTM Indentation Hardness of Rubber and Plastics by Means of a Durometer (D2240-68)

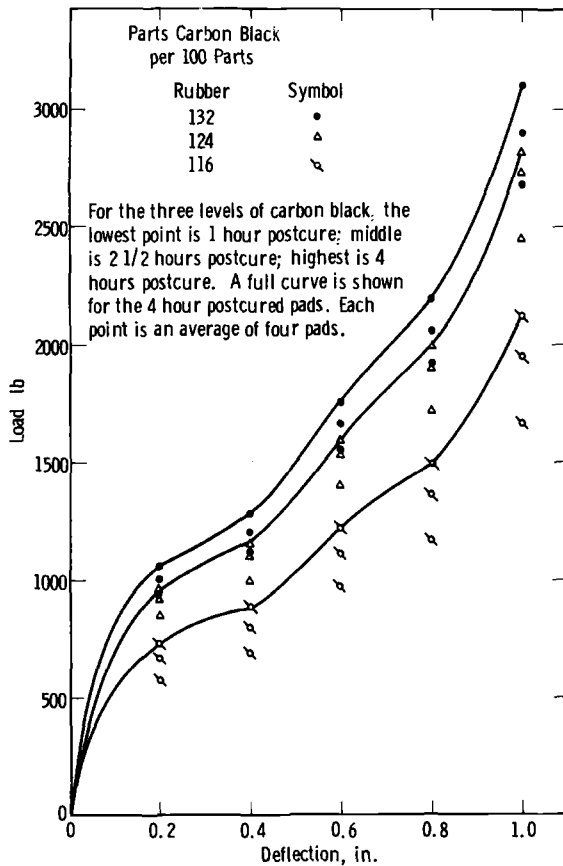


FIG. 2—C-D data for three carbon black loadings and three postcure times at 150 C in a laboratory oven.

Pads postcured 0, 3 and 6½ h at 150 C were selected to bracket what we thought would be the optimum to meet the required stiffening criteria. These were then evaluated in terms of static C-D stiffness before and after aging. The 6½-h postcure is also the practical maximum that can be accomplished in an 8-h shift. Sections of pads were used as the specimens since, during the development stage, little success was realized in correlating the basic mechanical properties of the rubber with actual tactical performance. This was primarily for two reasons; (1) an accurate stress analysis of the liner pad does not exist, largely because of its complexity, and (2) material molded to simple shapes such as sheets or slabs for the preparation of tension or compression specimens does not exhibit the same mechanical properties as the same material molded to the geometry of the liner pad. This is because of the cure conditions associated with the complicated and unique rubber stock flow patterns peculiar to the liner pad mold. This variation is further complicated by

production postcuring conditions where the time-temperature variations within the oven are considerable.

A photograph of a complete pad and a specimen made by sectioning a pad is shown in Fig. 3. Figure 4 shows the cross section of a pad in progressive stages of compression, illustrating the complex strain conditions inherent with the prebuckled strut type design.

About 80 duplicate tests were run in this evaluation requiring 20 full scale pads for each of the three postcures, namely 0, 3 and 6½ h at 150 C. All postcuring was done in the vendor's commercial oven where we measured the temperature variation with respect to time and location. The time refers to dwell time after the room temperature pads are loaded in the preheated oven. After postcuring the pads are allowed to cool under ambient room conditions.

The change in C-D stiffness of each specimen is defined as

$$\text{percent change} = \frac{\text{load after aging} - \text{precrush load}}{\text{precrush load}} \times 100.$$

Under production conditions each pad is qualification tested 24 to 120 h after the precrush (virgin) test. This C-D test is termed the test for record (TFR) and must fall within specified limits for the pad to be accepted. The load coordinate of the TFR test at any given deflection is always lower than for the precrush. This is because the recovery mechanism alone will not restore the rubber to the initial state in a reasonable time. The load or stress difference between successive compressive or extension cycles is associated with the normal "Mullins softening" of a filled vulcanizate[4]. On aging at room temperature, the near complete

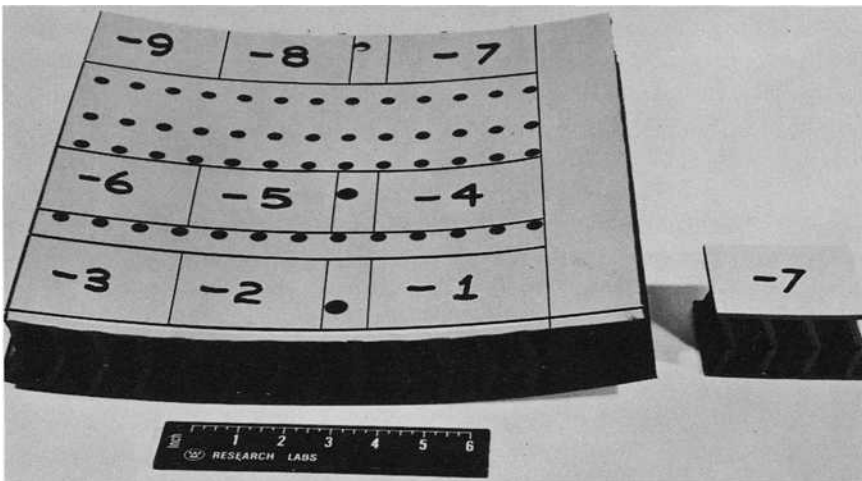


FIG. 3—A photograph of a full scale neoprene liner pad. Nine 4-strut specimens were cut from each pad as shown. The pin holes are from chaplet pins used to support the core bars and to locate and hold the Teflon sheet during molding.

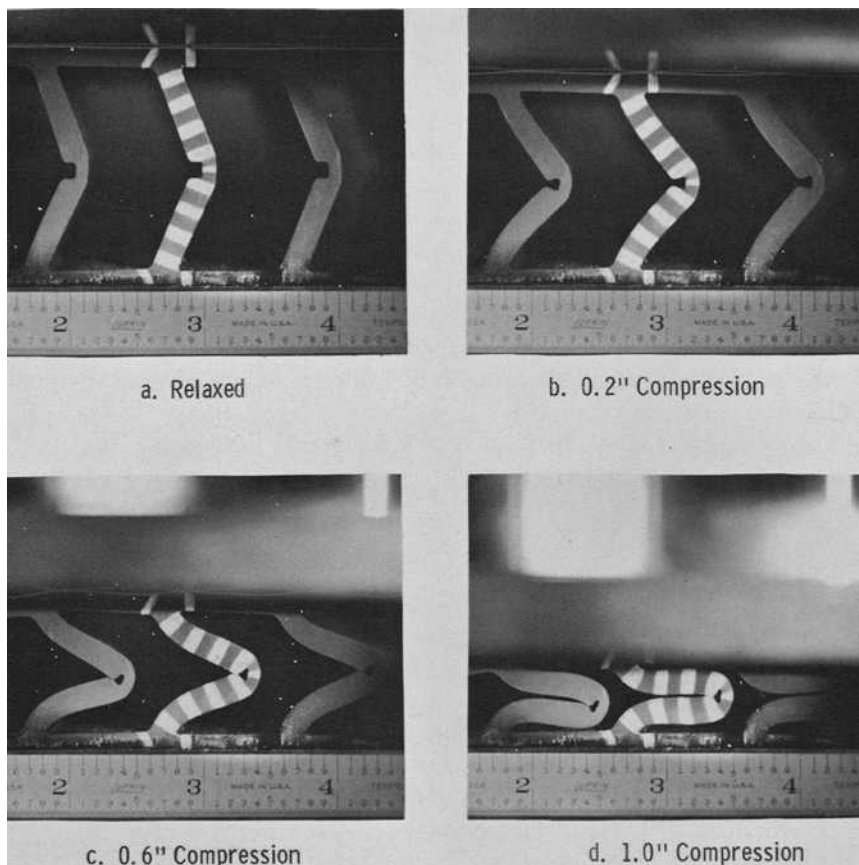


FIG. 4—Cross section of a neoprene liner pad in varying stages of compression.

recovery from the second cycle to initial or virgin values for a well cured vulcanizate should require a very long time and is associated with "Mullins recovery." In 1947, Mullins showed that for black filled natural rubber vulcanizates (~ 35 percent volume fraction carbon black) only about 15 percent recovery of load was attained between the first and second extensions of rubber strips after more than 100 h aging at room temperature. After 100 h at 100 C the recovery was about 90 percent. Although not specifically mentioned, it is obvious that Mullins worked with well cured vulcanizates since no increase above initial or virgin load values was recorded. This recovery mechanism is intrinsically included in the measured stiffening of the pad specimens.

Thus to predict life in actual service it was decided to use the change in C-D stiffness of the pad as the life determining parameter. Sections of pads of a given postcure history were randomly selected and assigned to aging temperatures ranging from 45 to 95 C. By using the precrush test

of each specimen as a base for comparison, each specimen serves as its own control and negates the pad-to-pad and within-pad stiffness variations.

Experimental Procedures

Elevated Temperature Aging in Air

All pads used in these experiments were molded in the same mold by the Republic Rubber Division, Aeroquip Corp., Youngstown, Ohio. Pads of a given postcure were cut into nine specimens and then randomly assigned to the five aging temperatures which were 55 to 95 C in 10 C increments for the 6½-h postcured pads and 45 to 85 C for the 0 and 3-h postcured pads. All specimens were precrushed in a known sequence on one day and tested for record in the same sequence on the following day. The test time interval was thus about 24 h for each specimen. The specimens were then placed in the aging ovens at temperatures just mentioned. Pairs of specimens were removed from the oven on a prearranged schedule, cooled for 4 to 24 h at room temperature and given the after-aging C-D test. Test intervals were adjusted as testing progressed and the aging rates developed. The same procedure was followed for each of the three postcures on a staggered schedule.

All C-D testing was accomplished in an Instron Universal Testing Machine at 2.0 in./min crosshead rate. The crown of the approximately 2.6 by 3.5 in.-specimens, due to the pad curvature, was ignored and flat Micarta test platens were used. Autographic records were taken for each C-D test and loads at 0.4 and 0.6 in.-deflections used to calculate stiffness changes. Fig. 5 shows that the precrush, TFR and after-aging C-D curves are all of similar shape, so data at these explicit deflections were used to characterize the overall C-D response.

Aging in a Nitrogen Atmosphere

An experiment conducted concurrently with the aging studies in air involved aging in nitrogen to eliminate oxidation effects. Specimens were sectioned from full scale pads as before, precrushed, tested for record, and then placed in 1-gal paint cans. Hoses connected a series of these cans to a regulated nitrogen cylinder. The cans were placed in a 70 C oven, aged up to 30 days and tested as previously described. A slight positive nitrogen pressure was maintained throughout the tests to purge the cans of air.

Crosslink Density Measurements

Crosslink density values were measured on portions of the pads before and after aging. The intent was to correlate the change in state of cure with the pad stiffness. As it turned out the measured molecular weight between crosslinks, $\bar{M}_c$, was found to be nearly invariant with any of the

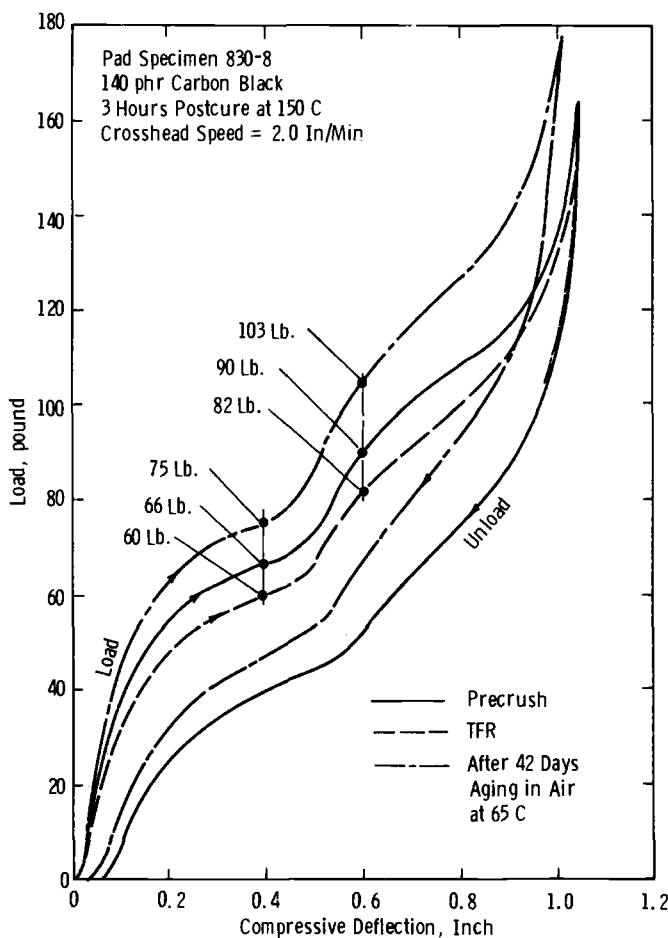


FIG. 5—Tracings of typical Instron C-D records. The shapes of all records are similar and the loads at 0.4 and 0.6-in. deflection have been used to characterize the entire curve.

thermal agings employed. We suspect that the techniques used are not sensitive enough to detect the changes that occur.

Two techniques were used to measure crosslink density. One involves measuring the load-deflection slope of a solvent swollen disk or stack of disks cut from the aged rubber[5]. The slope is then substituted in the equation

$$\frac{v_e}{V} = \frac{h_o S}{3 A_o R T}$$

and the crosslink density, v_e/V , is computed. The symbols in the equation are defined as follows:

$\frac{v_e}{V}$ = crosslink density, mol/cc,

h_o = height of unswollen test specimen, cm,

- S = slope of load deflection curve, g/mil,
 A_o = area of unswollen specimen, cm²,
 R = gas constant, (cal)/(deg K) (g-mol), and
 T = temperature, deg K.

A multiplier, 394 mils/cm, is necessary in the numerator of the above equation to obtain v_e/V in the proper units.

The second technique involves the equilibrium swelling of the rubber vulcanizate and the relation of the swelling ratio, $v_2 = \frac{1}{g_m}$ to the molecular weight between crosslinks, $\bar{M}_c$ [6]. The relationship used is

$$\bar{M}_c = \frac{d_1 V_1}{(v_2)^{5/3} \left(\frac{1}{2} - \mu \right)}$$

where

- d_1 = density of rubber sample, g/cc.,
 V_1 = molar volume of the swelling solvent, cc/mol (88.6 for benzene and 97.1 for carbon tetrachloride),
 v_2 = swelling ratio, $\frac{1}{g_m}$, where $g_m = \frac{\text{volume swollen polymer}}{\text{volume dry polymer}}$, and
 μ = polymer interaction coefficient, 0.26 for neoprene-benzene and 0.414 for neoprene-carbon tetrachloride[6].

The relationship between the molecular weight between crosslinks, sample density and the crosslink density

$$\bar{M}_c = \frac{d_1}{\frac{v_e}{V}}$$

The solvents used in these techniques were toluene, benzene, and carbon tetrachloride. The specimens for the first technique were $\frac{5}{8}$ -in. diameter disks cut from the face sheet between the struts. The specimens used for swelling ratio measurements were of random shape and taken from the notch area of the pad strut.

Results

A Two-Step Approach to Life Prediction Based on the Arrhenius Relationship

The aging data plotted as percent change from precrush load versus (days aged)^{1/2} for a given aging temperature suggests a straight line relationship. The data for the unpostcured specimens is presented in Fig. 6. Data for the other postcures is similar except the slopes of the least squares fit lines decreases with increased postcure time. A summary of all aging data is listed in Table 2. Point estimates on the least

TABLE 2—Summary of aging data for neoprene pad specimens.

Postcure (hours)	Carbon Black (phr)	Aging Temperature (deg C)	Point Estimate on Least Squares Fit Line - Time Required to Stiffen the Indicated Amount (days)				R ² Correlation ^a Coefficient
			0%	5%	10%	15%	
0 ^b	140	45	2.58	11.1	25.5	45.9	0.82
		55	1.18	4.79	10.8	19.3	0.95
		65	0.51	1.97	4.37	7.71	0.89
		75	0.34	0.93	1.83	3.02	0.98
		85	0.11	0.53	1.25	2.29	0.90
3	140	45	20.1	65.7	137.4	235.3	0.86
		55	9.02	25.7	50.8	84.5	0.93
		65	6.00	14.6	26.9	42.9	0.97
		75	2.37	4.82	8.13	12.3	0.98
		85	0.85	2.04	3.75	5.98	0.99
6 1/2	124	55	53.4	100.7	162.8	239.9	0.94
		65	16.9	34.2	57.5	86.9	0.97
		75	7.37	12.9	20.1	28.7	0.98
		85	4.61	7.92	12.1	17.2	0.92
		95	2.14	3.38	4.88	6.67	0.97

a A measure of how well the data fit the regression line on "percent change from precrush" versus "(days)^{1/2}" coordinates.

b Data for the unpostcured pads are presented graphically in Figure 6.

squares fit lines at 0, 5, 10 and 15-percent load increase from precrush are then used to graphically construct the Arrhenius relationship[7].

$$k = Ae^{\frac{-E}{RT}}$$

or

$$\log k = \left(\frac{-E}{2.303R} \right) \frac{1}{T} + A$$

where

k = a time increment, days, etc.,

A = constant,

E = energy of activation, K cal/mol,

R = gas constant, 1.987 (cal)/(deg K) (g-mol), and

T = temperature, deg K.

Figures 7, 8, and 9 show this graphical construction for the three postcures and the least squares fit lines that are extrapolated to predict life at 25 C.

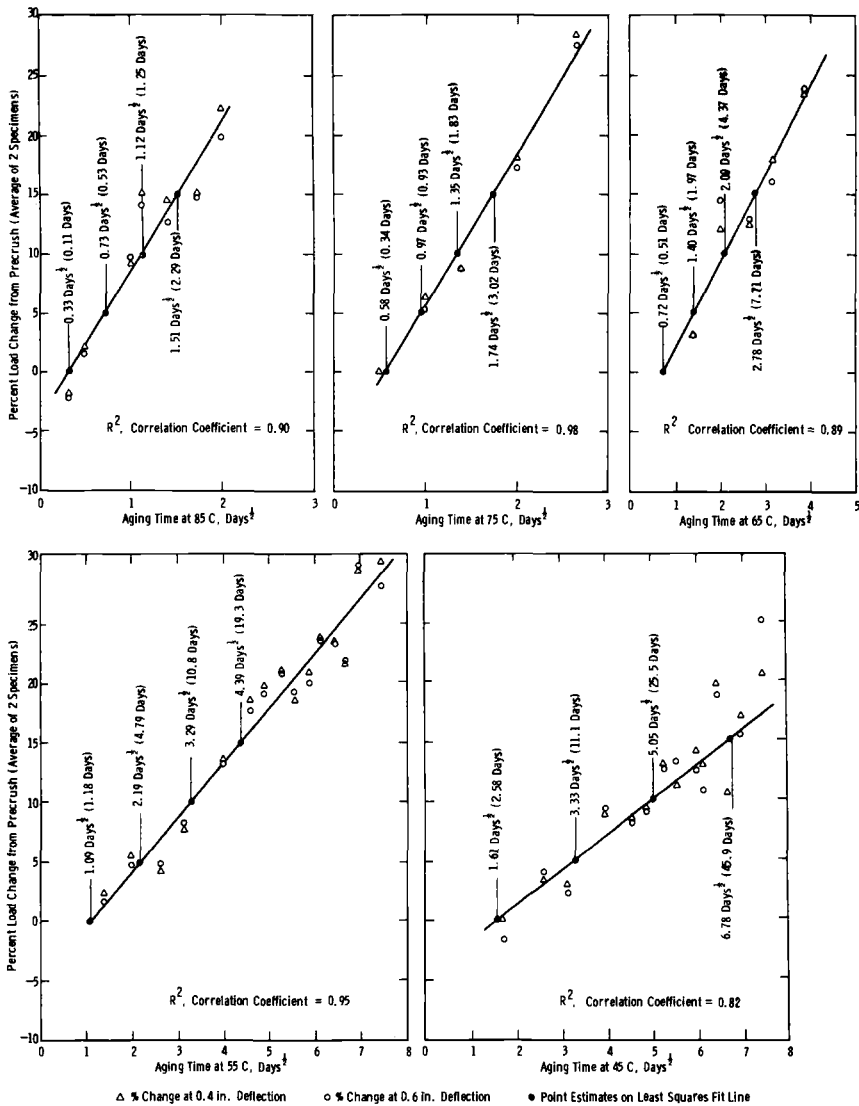


FIG. 6—Stiffening data for unpostcured neoprene pad specimens molded from stock with 140-phr carbon black. Aging temperatures range from 45 to 85 C.

A More Rigorous Statistical Analysis for Life Prediction

The above two-step method does not allow construction of exact confidence limits on the predicted life values. To overcome this shortcoming, Dr. G. M. Jouris of the Westinghouse Research Mathematics Dept. chose a model of the form

$$P = a_1 + a_2 \log_{10} A + a_3 T'$$

where

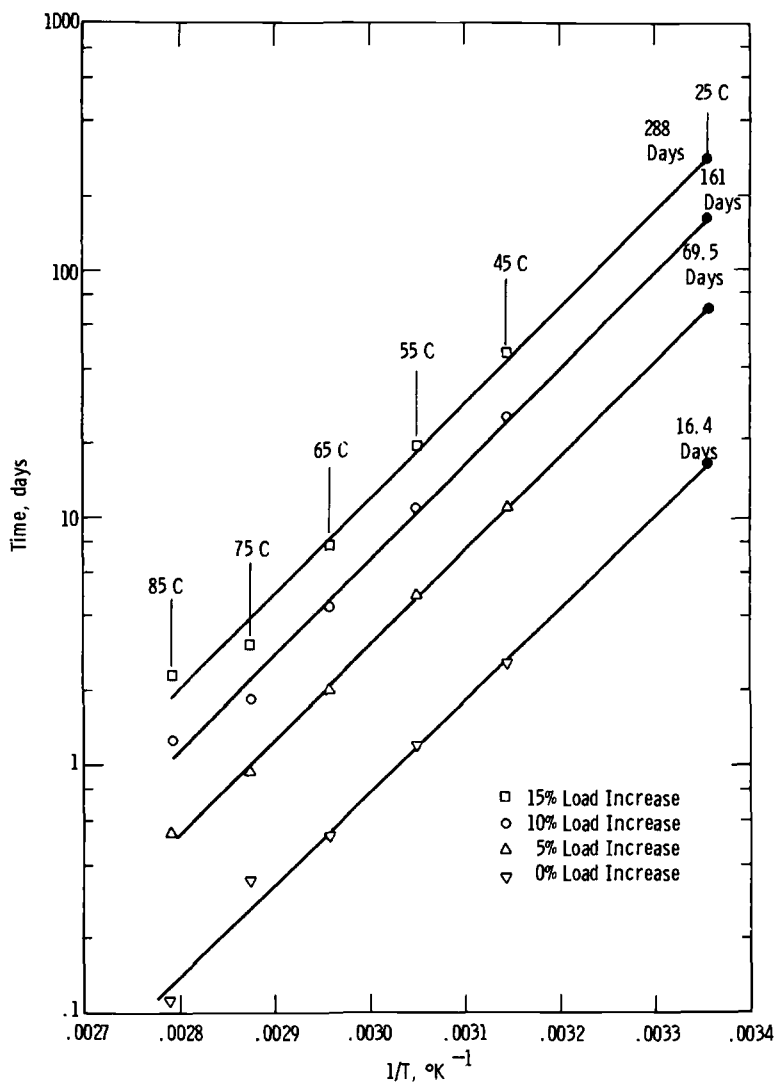


FIG. 7—Aging data plotted according to the Arrhenius relationship. Extrapolation to 25 C (room temperature) is for specimens molded from 140-phr carbon black stock and given no postcure.

P = percent increase above precrush,

a_i = constants,

A = aging or life time, days, and

T' = inverse of temperature, deg K⁻¹.

This model is, in a sense, a combination of the models used in the two-step approach. In Table 3 the results of this analysis are summarized and compared to the two-step approach.

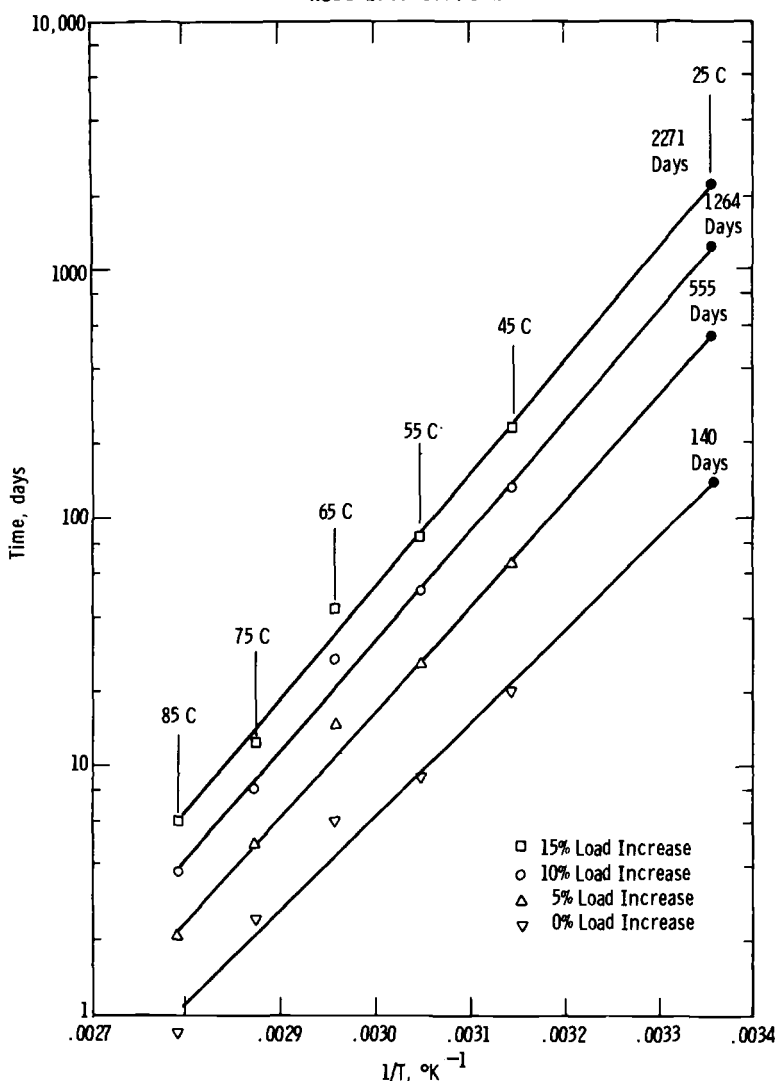


FIG. 8—Aging data plotted according to the Arrhenius relationship. Extrapolation to 25 C (room temperature) is for specimens molded from 150-phr carbon black stock and postcured for 3 h at 150 C.

Crosslink Density and Mechanical Properties

The crosslink density and mechanical property measurements are summarized in Table 4. The column-to-column difference between $\bar{M}_c$ values obtained by the two techniques are largely due to differences in solvent power of the solutions used. However, differences within the columns of the table reflect the change in measured crosslink density as a function of the postcures and thermal agings listed.

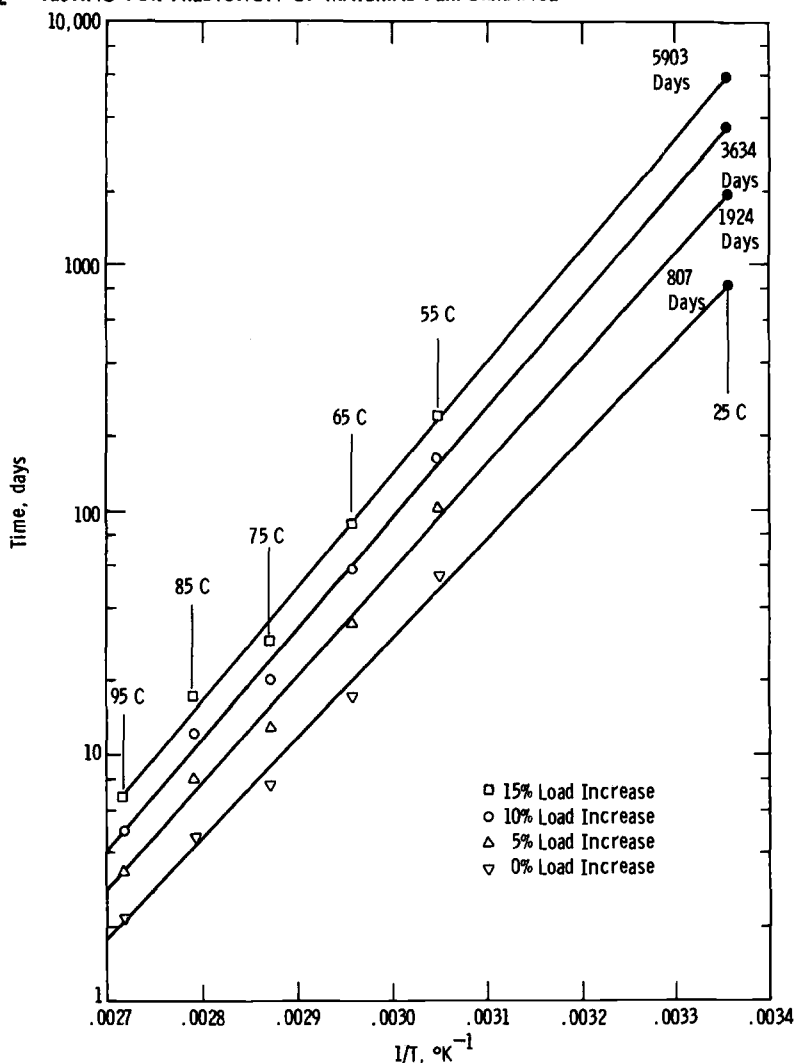


FIG. 9—Aging data plotted according to the Arrhenius relationship. Extrapolation to 25 C (room temperature) is for specimens molded from 124-phr carbon black stock and postcured for 6½ h at 150 C.

Discussion

The useful service life of the neoprene pads has been defined in terms of the increase in static stiffness. We have reduced the data for 0, 5, 10, and 15 percent increase above the precrush value. The concept of "zero percent increase" may seem trivial but it involves the recovery mechanism mentioned previously. A pad has to actually stiffen to duplicate the precrush or virgin C-D curve since recovery is incomplete.

The two methods of data analysis give predictions that are in good agreement. This is to be expected as the models are quite similar. Each approach has its advantages and disadvantages. Calculations for the two-

TABLE 3—Life predictions from the model $P = a_1 + a_2 \log_{10} A + a_3 T$ and comparisons with the two-step approach.

A. Life predictions based on a statistical analysis using the above mathematical model									
No Postcure					3 Hour Postcure				
Fitted Model: $P = 219.01 + 18.33 \log_{10} A - 7.40T$					Fitted Model: $P = 175.90 + 13.41 \log_{10} A - 6.16T$				
$R^2 = 0.89$ $s = 3.07$					$R^2 = 0.84$ $s = 3.54$				
P	$\hat{A}^c$	L_{90}^d	L_{95}^d		P	$\hat{A}$	L_{90}	L_{95}	
0%	38.8 days	21.9 days	18.5 days		0%	196.2 days	82.6 days	63.9 days	
5	72.7	41.6	35.2		5	462.9	195.9	152.2	
10	136.3	78.5	66.6		10	1092.1	462.0	359.8	
15	295.5	147.7	125.7		15	2576.7	1083.9	844.3	
Fitted Model: $P = 201.92 + 1706 \log_{10} A - 6.79T$					Fitted Model: $P = 166.77 + 12.99 \log_{10} A - 5.85T$				
$R^2 = 0.85$ $s = 3.55$					$R^2 = 0.84$ $s = 3.44$				
P	$\hat{A}$	L_{90}	L_{95}		P	$\hat{A}$	L_{90}	L_{95}	
0%	33.0 days	16.2 days	13.0 days		0%	187.1 days	78.5 days	60.8 days	
5	64.8	32.3	26.1		5	454.0	191.7	148.9	
10	127.2	64.0	52.2		10	1101.5	465.0	361.9	
15	249.7	126.5	103.5		15	2672.7	1121.5	873.1	
Fitted Model: $P = 198.14 + 16.26 \log_{10} A - 7.36T$					Fitted Model: $P = 992.4$ days				
$R^2 = 0.82$ $s = 3.76$					$R^2 = 0.81$ $s = 3.79$				
P	$\hat{A}$	L_{90}	L_{95}		P	$\hat{A}$	L_{90}	L_{95}	
0%	1042.2 days	485.2 days	387.3 days		0%	992.4 days	448.7 days	355.0 days	
5	2076.8	968.8	775.5		5	2014.4	913.2	724.7	
10	4138.6	1925.4	1543.0		10	4089.0	1849.0	1469.4	
15	8247.3	3808.9	3051.6		15	8300.0	3724.8	2959.9	
B. Comparison of Prediction Methods									
P	$\hat{A}$ (Avg from part A above)	$\hat{A}$ (from Fig. 7)	% difference		P	$\hat{A}$ (Avg from part A above)	$\hat{A}$ (from Fig. 8)	% difference	
0%	35.9 days	16.4 days	-54%		0%	191.6 days	140 days	-27%	
5	68.8	69.5	+1		5	458.4	555	+21	
10	131.8	161	+22		10	1096.8	1264	+15	
15	252.6	288	+14		15	2624.7	2271	-13	
C. Comparison of Prediction Methods									
P	$\hat{A}$ (Avg from part A above)	$\hat{A}$ (from Fig. 9)	% difference		P	$\hat{A}$ (Avg from part A above)	$\hat{A}$ (from Fig. 9)	% difference	
0%	1017.3 days	807 days	-21%		0%	1017.3 days	807 days	-21%	
5	2045.6	1924	-6		5	2045.6	1924	-6	
10	4113.8	3634	-12		10	4113.8	3634	-12	
15	8273.6	5903	-29		15	8273.6	5903	-29	

a R^2 = multiple correlation coefficient.

b s = standard deviation of the residuals.

c $\hat{A}$ = point estimate of the pad life.

d L_{90} and L_{95} are the lower 90 and 95% confidence limits respectively.

TABLE 4—Range of molecular weights between crosslinks and mechanical properties of neoprene specimens aged at elevated temperatures in air.

Postcure (hours at 150°C)	Aging Conditions	Molecular Weight Between Crosslinks, $\overline{M}_c$ (g / mol)			Tensile ^d Strength (psi)	100% ^d Modulus (psi)	Elongation ^d (%)
		a	b	c			
0	Control	1800	5600	• • •	1680	1025	240
	3-24 Days at 45°C	1600-1900		3000-3800	1545-1695	1015-1185	210-240
	2-24 Days at 55°C	1800-1975	• • •	2600-3350	1410-1890	930-1215	200-250
	2-24 Days at 65°C	1700-1900	4125-6000	2600-3600	1470-1745	1065-1350	135-235
	0, 25-21 Days at 75°C	1800-2000	• • •	2250-3000	1465-1775	995-1280	185-255
	0, 25-18 Days at 85°C	1600-1900	5050-5300	2500	1515-1840	880-1500	120-300
3	Control	1500	4600-5500	• • •	1465	1240	160
	8-28 Days at 45°C	1600-1700	4250-4350	• • •	1530	1320	150
	3-28 Days at 55°C	1600-1700	4250-4450	2200-2550	1520-1880	1280-1565	130-200
	2-16 Days at 65°C	1600-1750	• • •	1700-2500	1575-1800	1300-1530	135-185
	0, 25-28 Days at 75°C	1245-1700	4950-5000	2200-2500	1465-1795	1290-1705	90-185
	0, 12-18 Days at 85°C	1600-1825	4700-5200	2100-2700	1485-1825	1365-1545	110-175
6 1/2	Control	• • •	• • •	2270	1275-1490	1360	120-170
	5-31 Days at 70°C	• • •	• • •	1860-2090	1525-1690	1430-1660	110-130

a Determined from swelling ratio after swelling in benzene.

b Determined from swelling ratio after swelling in carbon tetrachloride.

c Determined from load-deflection slope after swelling in toluene.

d ASTM Tension Test for Vulcanized Rubber (D412-68).

step approach can be easily done on a programmable desk top calculator. The fact that a square root time transformation tends to linearize the air aging data may give insight into the mechanisms involved. Fig. 10, data from the 70 C nitrogen aging tests, illustrates that oxidation is part of the aging process. The straight lines on the figure are the averages from 65 and 75 C air aging tests for the three postcures. The nitrogen aged data points are generally below their respective air aged lines indicating that oxidation occurs in air aging. Another advantage for the two-step ap-

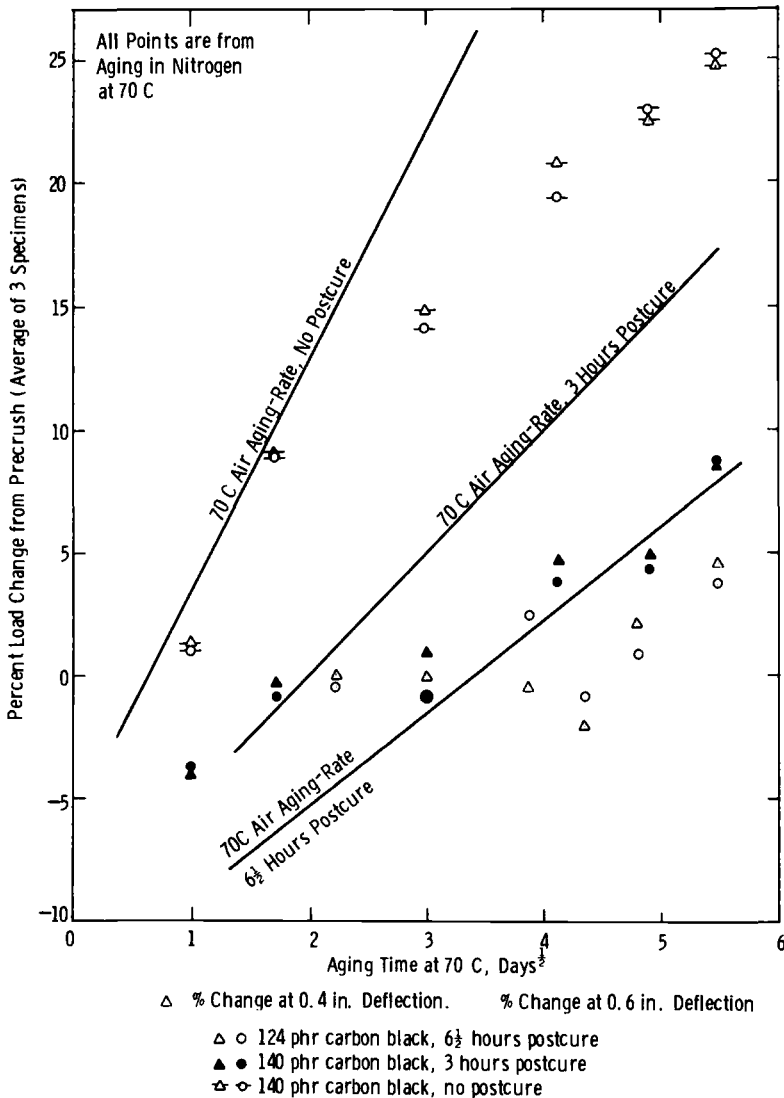


FIG. 10—Comparison of stiffening rates for neoprene pad specimens aged in air or nitrogen at 70 C.

proach is that the data plotted in the Arrhenius form allows one to calculate the energy of activation from the relationship

$$E = -2.303R (\text{slope})$$

These values may also aid in determining the mechanisms of the aging process. At this time, however, we are concerned with predicting the useful service life of the pads and have not spent much time trying to define the process.

The principle advantage of the one-step approach is that confidence limits on the prediction can be calculated. Note in Table 3 that the interval from the 90 percent lower confidence limit, L_{90} , to the point estimate, $\bar{A}$, for a given percent increase, P , usually overlaps the interval for the next highest P . For example, the 3-h postcure data shows that the time required to stiffen zero percent is from 82.6 to 196.2 days,—to stiffen five percent the interval is 195.9 to 462.9 days, etc. Naturally the overlap is even greater if the lower 95 percent confidence limit, L_{95} , is used in place of L_{90} . With a few exceptions the point estimates from the two-step approach fall within the L_{90} to $\bar{A}$ intervals. The exceptions are on the high side for which limits were not calculated.

In the one-step approach the model was fit separately to the 0.4 and 0.6-in. deflection data, whereas in the two-step approach the data were lumped together. Practically speaking, the results are so close that the difference is insignificant.

By either method of analysis, the benefit of postcuring is clearly demonstrated. Based on the two-step analysis pads with 6½-h postcure at 150 C require about 23 times as long to stiffen ten percent above pre-crush as those with no postcure. Pads with 3-h postcure at 150 C will take about eight times as long to stiffen a like amount. These predictions are for service life in air at 25 C. We are currently aging specimens at room temperature to verify these predictions. When completed, these specimens will have been aged for up to 90 weeks.

The use of crosslink density and mechanical properties data to predict life is not practical for the neoprene pads. Table 4 shows that in a gross sense at least, the $\bar{M}_c$ values decrease with increased postcure at 150 C. However, aging at temperatures up to 95 C did not produce enough $\bar{M}_c$ variation to be useful. It is also shown that tensile strength and elongation decrease with postcure and 100-percent modulus increases. Again the variation with aging is not as sensitive as changes in C-D stiffness.

Conclusions

The data in this paper clearly illustrate that increasing the postcure reduces the tendency for pads to stiffen with time. The useful service life is considerably increased at no significant increase in original pad cost.

Two tools are available to predict this life—each has particular ad-

vantages. To summarize, the two-step approach may give insight into the aging mechanisms involved. Also the calculations are easily performed with a desk top calculator. The more rigorous one-step treatment offers the advantage of providing confidence limits on the estimates of pad life.

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Fundamentals of Radiation Effects Testing

REFERENCE: Sinclair, E. E., Simmons, J. M., and Zwilsky, K. M., "**Fundamentals of Radiation Effects Testing**," *Testing for Prediction of Material Performance in Structures and Components*, ASTM STP 515, American Society for Testing and Materials, 1972, pp. 198–228.

ABSTRACT: The basic effects of irradiation, such as transmutation reactions, hardening, void formation and high temperature embrittlement are reviewed. Testing for these effects takes the form of pre-post testing as well as in-reactor observations. Pre-post testing consists of measuring a property out-of-reactor, and measuring the same property after the material has been exposed to the neutron environment. Although changes in properties of a material can be determined before and after irradiation, the influence of a neutron flux on rate processes such as creep must be determined in-pile.

Characterization of the reactor environment including neutron dosimetry and temperature measurements are discussed, and examples of predictive testing of materials for reactor fuels, cladding, components and pressure vessels are reviewed.

KEY WORDS: predictions, irradiation, nuclear reactor, nuclear energy, hardening (systems), voids, embrittlement, radiation dosimetry, nuclear fuels, cladding, pressure vessels, damage

Predictive testing for irradiation effects has been an integral part of reactor development since the earliest design and construction of nuclear test and power reactors. Since little was known of the effects of neutrons on materials, reactor materials had to be qualified on a step by step basis, as reactor lifetime and neutron exposures continually increased.

When nuclear fission takes place in a reactor, high energy fission fragments, beta and gamma radiation, and neutrons of varying energies are produced. The fission fragments are primarily responsible for lattice damage to the fuel body, since their energy is given up in the space of only a few micrometers. The fuel may be damaged quite heavily and such damage may be the limiting factor in the life of a fuel element. However, the damage to be discussed in this paper is that caused by neutrons whose range, in contrast to that of the fission fragments, is very great. Neutrons affect every component within the reactor vessel and although

¹ U. S. Atomic Energy Commission, Division of Reactor Development and Technology, Washington, D. C. 20545.

the rate of damage may be relatively low, the total damage sustained by a component during the lifetime of the reactor may be significant.

Although the importance of radiation effects testing was recognized early, testing techniques have developed as an art rather than a science. This is due in part to the differences in test reactors as well as to the wide scope of test parameters of interest. Testing space varies according to the type of reactor, position in the reactor, accessibility for instrumentation, neutronic considerations, reactor environment and general support available to the experimenter. The type of information that is required varies from physical and mechanical properties of the fuel and cladding to integral performance of fuel pins and subassemblies, as well as the performance of a variety of structural components. It is of little surprise, therefore, that few standard forms of capsules for radiation testing have emerged over the years.

The purpose of this paper is to underline the importance of predictive testing by citing a number of case histories in which such testing, or the lack of it, played a decisive role in the performance of a reactor material or component. Prior to this discussion, however, a general outline of the effects of neutron radiation on structural materials is presented, followed by a section dealing with the considerations that influence experiment design and a section on the characterization of the reactor environment. Since radiation effects testing is expensive and limited by suitable reactor space, all of these factors need to be considered carefully to obtain the maximum benefit from each individual experiment.

Effects of Irradiation on Structural Materials

The principle contribution to irradiation effects in metals comes from neutron irradiation. Beta and gamma radiation produce primarily electronic excitation and ionization, which cause little permanent damage in metals. However, it should be pointed out that electronic excitation produced (especially by gamma rays) results in vibrational energy of atomic nuclei and therefore appears in the form of heat. Thus, even though the direct effect of such radiation is small, secondary effects due to heating and therefore annealing of neutron induced damage may be encountered.

Neutron irradiation produces two basic types of damage. The first is atomic displacement due to elastic collision of neutrons with the atoms in the metal lattice. The second is transmutation of the constituent atoms due to capture of both thermal and fast neutrons. These two types of damage and their effects are described below.

Displacement Damage

The kinetic energy needed to permanently displace an atom from its lattice site is in the range of 25 eV. The average number of displacements as a function of neutron energy for pure iron is shown in Fig. 1.

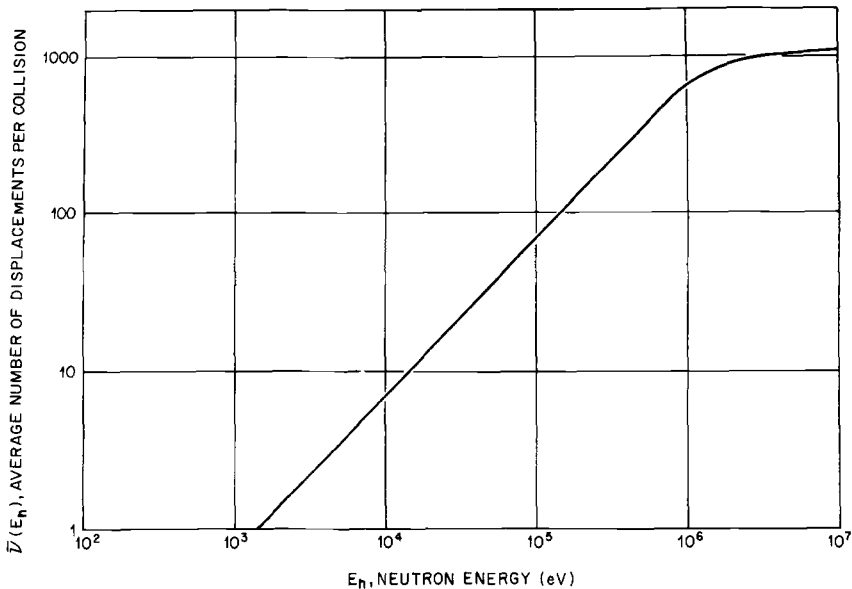


FIG. 1—Average number of atom displacement as a function of neutron energy in iron[1].

At low neutron energies the energy transfer between the neutron and the atom is insufficient to cause permanent displacement. At intermediate neutron energies the number of gross displacements is approximately proportional to the neutron energy, whereas at high neutron energies (greater than about 1 MeV) the displacements saturate rapidly due to ionization processes[1].

An atom which receives sufficient energy to be displaced from its site is called a primary knock-on atom (PKA). Such an atom travels through the lattice and strikes other atoms transferring energy to them and likewise displacing them from their site. In general, each PKA initiates a cascade of interdependent atomic collisions, and this cascade produces a distribution of vacancies and interstitials which is called a primary damage state or a primary defect distribution state[2]. Two classes of defects are introduced into the structure by the collision of neutrons and atoms. First are the point defects including vacant lattice sites where an atom used to be, or interstitial atoms which are individual atoms displaced from their lattice position. Second are the cluster defects consisting of either interstitials or vacancies. Since reactor irradiations are usually carried out at temperatures where one or more of the defects is mobile, the primary damage state is highly unstable and dynamic recovery accounts for a partial rearrangement of the lattice. The behavior of such defects, their migration, growth, annihilation, interaction with other defects and precipitates, and annealing determines the physical and me-

chanical properties which must be established to predict the performance of a given material.

Transmutation Reactions

Instead of elastic collisions between neutrons and nuclei, a neutron may be captured by the nucleus of an atom, producing either a new isotope of the same species or a chemically different species. Transmutations generally occur more readily with neutrons of low energy, but do occur with neutrons of all energies and in all elements[3]. The most damaging reactions to metal properties are those that result in gaseous species, especially helium. Some typical reactions are listed in Table 1.

TABLE 1—*Examples of transmutation reactions* [8].

Helium Atom Formation					
$B^{10} (n, \alpha) Li^7$		$Cr^{53} (n, \alpha) Ti^{50}$			
$N^{14} (n, \alpha) B^{11}$		$Ni^{58} (n, \alpha) Fe^{55}$			
$Fe^{54} (n, \alpha) Cr^{51}$		$S^{32} (n, \alpha) Si^{29}$			
Hydrogen Atom Formation					
$Fe^{54} (n, p) Mn^{54}$					
$Ni^{58} (n, p) Co^{58}$					
$Al^{27} (n, p) Mg^{27}$					
Metallic Atom Formation					
$W^{187} (n, \gamma) W^{188}$	69 days	Re^{188}	17 h	Os^{188}	
$S^{34} (n, \gamma) S^{35}$	86.7 days	Cl^{38}			
$Cr^{54} (n, \gamma) Cr^{55}$	3.5 min	Mn^{55}			

In the case of (n, α) and (n, p) reactions, transmutation is the direct result of neutron capture whereas for (n, γ) reactions, transmutation is the result of neutron capture followed by beta decay.

The usual source of helium atoms in structural materials has been postulated to be the $B^{10} (n, \alpha)$ reaction for which the cross section increases with decreasing neutron energy. Materials containing boron thus have a high helium content after exposure in a thermal neutron environment. High energy neutrons, essentially those above 1 MeV, also lead to helium formation by (n, α) reactions, some of which are illustrated in Table 1. These reactions take place in a fast neutron environment so that for example, helium is produced in the stainless steel core components of a fast reactor.

More recently it has been suggested that in mixed fast-neutron and thermal-neutron fluxes, helium production is much larger than currently assumed[4]. A two stage process has been postulated in which helium is generated by thermal neutron reaction with ^{58}Co formed from the (n, p) reaction with Ni. Other paths involving ^{59}Ni and ^{56}Fe have also been suggested[5], and it seems likely that a complex series of reactions leads

to observed helium contents that are higher than predicted from single reaction processes.

The importance of hydrogen produced by (n,p) reactions is not clear. Since it diffuses rapidly at elevated temperatures, it is generally assumed not to be a problem. Small quantities of manganese, chromium, titanium, iron, and other transmutation products are assumed to have little effect *per se* on mechanical properties, since these elements are usually present as impurities already. However, in a few reactions the quantity of transmuted atoms is such that the compositional change becomes apparent in the properties of the alloys. One such reaction is the (n,γ) reaction for tungsten which will transmute to rhenium and rhenium to osmium. Solid solution strengthening of tungsten with decreasing ductility has been observed as a result of rhenium formation[3]. Of even greater importance than the change in mechanical properties is the change in emf of tungsten base thermocouples with increasing neutron exposure as illustrated later on.

Effects on Mechanical Properties

The changes in mechanical properties due to irradiation are a function of irradiation temperature, neutron fluence (flux integrated over the time of irradiation), and neutron spectrum. At low temperatures ($T < 0.4 T_m$ where T_m equals the melting temperature in absolute degrees), the macroscopic effects are an increase in yield strength and reduction of work hardening coefficient and uniform strain. Figures 2 and 3 illustrate these effects. Figure 2 presents a series of tensile curves for 304 stainless steel showing increasing yield and tensile strength and decreasing ductility with neutron fluence[6]. Figure 3 illustrates additional features often observed in irradiated material, such as a greater increase in yield than in tensile stress, decrease in strain hardening as the yield stress approaches the tensile stress, greater total strain than uniform strain and finally saturation of these effects with fluence. This saturation, although clearly shown for the particular experimental conditions described in Fig. 3, has not been demonstrated over a wide range of spectra and test conditions, and further work is required to establish whether it is valid under all conditions.

At intermediate irradiation temperatures ($0.4 T_m < T < 0.55 T_m$) the damage to metals consists primarily of voids and interstitial loops. Voids that are formed are on the order of several hundred angstroms in size and lead to a density decrease of the metal, which at high fluences can reach several percent. Examples of void formation are discussed in a subsequent section. Since the phenomenon was discovered recently, it makes one of the fascinating stories of predictive testing in progress at the present time. The overall effects besides swelling, which has perhaps the most important impact on design, is an increase in yield strength and

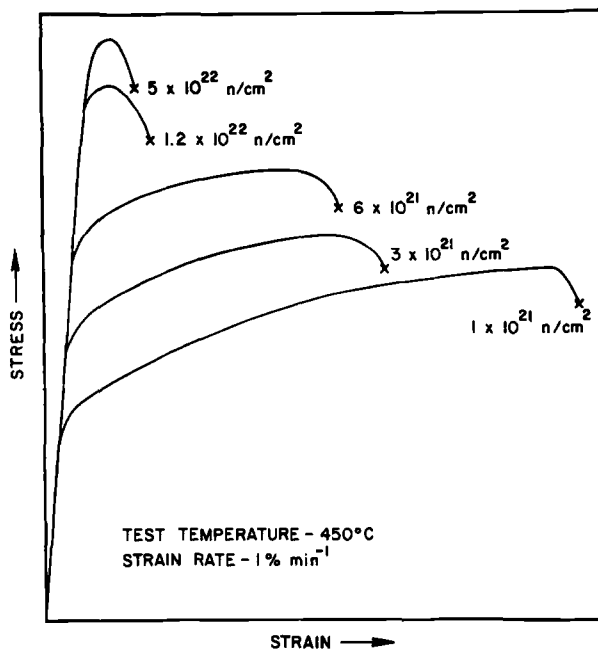


FIG. 2—Stress-strain curves for type 304 stainless steel irradiated at 371–460 C[6].

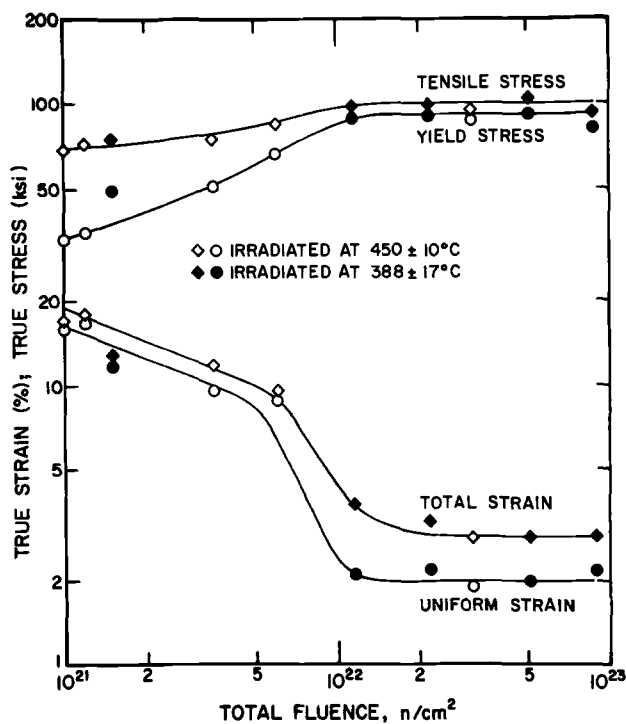


FIG. 3—Tensile properties at 450 C as a function of neutron fluence for type 304 stainless steel[6].

decrease in ductility. (A technique has been developed recently that predicts postirradiation tensile properties as a function of void diameter and volume change resulting from void formation[7]. This is one of the early attempts to correlate and predict mechanical properties from electron microscope observations.)

At high irradiation or test temperatures or both ($T > 0.55T_m$), strength properties of metals are not affected by irradiation. However, ductility is severely reduced, related to the helium bubbles formed by precipitation of helium atoms produced by (n, α) reactions. Figure 4 shows the effect of irradiation on tensile ductility[8]. The uniform elongation of the irradiated 347 stainless steel is 16 percent at 525 C, and drops rapidly to 1.8 percent at 575 C. Annealing of the material recovers the ductility at 500 C and below to a value close to that of the unirradiated comparison, an indication that the low temperature hardening is annealed out. However at temperatures greater than 500 C ductility is not recovered, indicative that the helium damage is not annealed out[8].

The effect of irradiation on stress-rupture properties is illustrated in Fig. 5. The rupture life of all irradiated specimens was reduced when compared to the unirradiated control material. At 538 C the substantial loss in rupture life was primarily a result of an increased creep rate, whereas at test temperatures of 593 C, 649 C and 760 C, the losses in rupture life resulted from reduced ductility due to helium embrittlement[9].

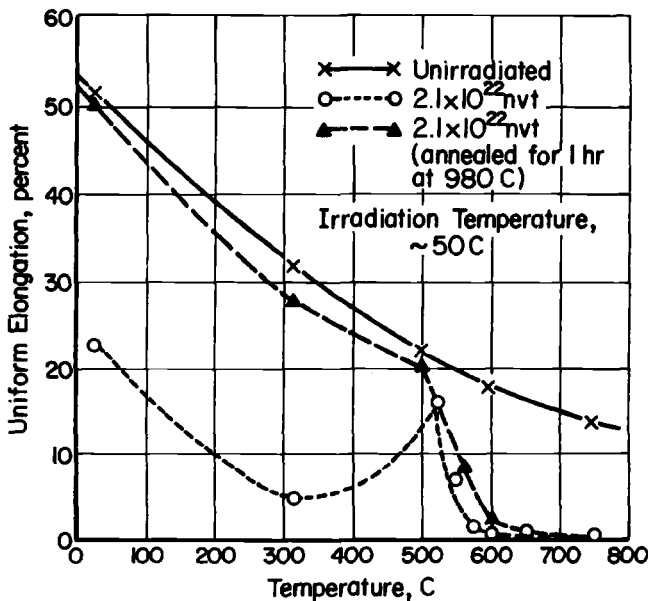


FIG. 4—Tensile ductility as a function of temperature for type 347 stainless steel unirradiated and irradiated at 50 C[8].

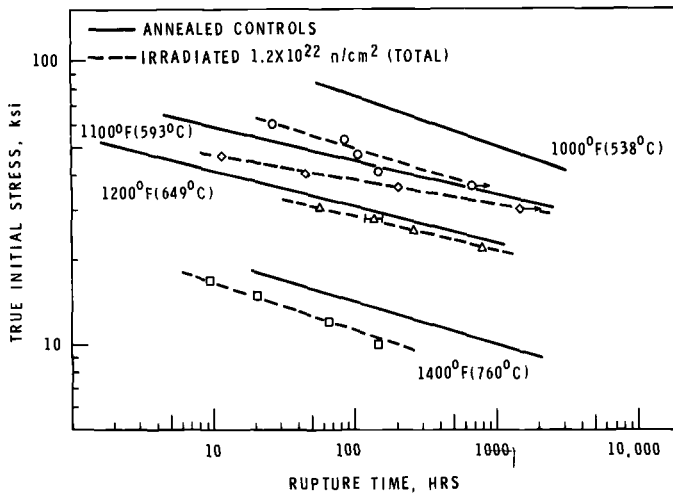


FIG. 5—Stress-rupture data for type 316 stainless steel unirradiated and irradiated at 482 C[9].

Experiment Design Considerations

Three symposia on irradiation effects testing have been cosponsored by the Atomic Energy Commission in the past eight years[10–12]. The three volumes resulting from the meetings contain a wealth of information on all aspects of capsule design, and should be consulted on the specifics of how to test in a reactor environment. The purpose of this section is to outline some of the general factors that need to be considered in the design of experiments.

Irradiation Capsules

Testing of materials is usually done in capsules for a number of reasons. Foremost among these is to keep the specimens separate from the reactor coolant, since unwanted reactions may contaminate both the specimen as well as the coolant which often has a carefully balanced chemistry. The use of capsules also assures that specimens fit into the reactor, and permits establishing a special test environment such as a high purity inert gas in cases where this is desired.

Capsules are either completely sealed and thus do not permit control during irradiation or else contain leads that permit external control during the experiment. Uninstrumented capsules lend themselves to the irradiation of many specimens in the shortest period of time and at a more moderate cost, whereas instrumented capsules are more difficult to operate because of the complex equipment inside and the necessary penetrations in the primary containment needed for instrumentation leads to the outside of the reactor. On the other hand, a direct reading of temperature or strain while the specimen is undergoing irradiation may be

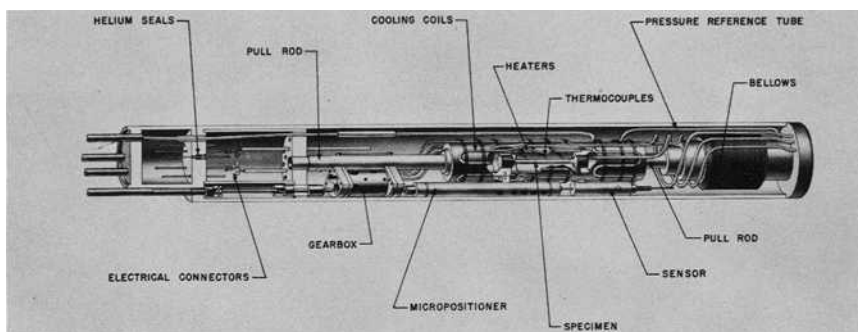


FIG. 6—Illustration of in-reactor creep capsule[13].

extremely important to the understanding of a particular phenomenon and may warrant the expense and sacrifice in reactor space.

A good example of the value of an in-reactor experiment is the question of how creep is affected by a neutron flux. Figure 6 shows a capsule used to measure in-pile creep. It can be surmised that a capsule of this type, although not necessarily more difficult to operate than other in-reactor experiments, is far more difficult to run than simply encapsulating a specimen followed by testing out-of-pile later on. Description of the details of operating this capsule can be found in Refs 13 and 14. The type of experimental results are illustrated in Fig. 7. Temperatures and

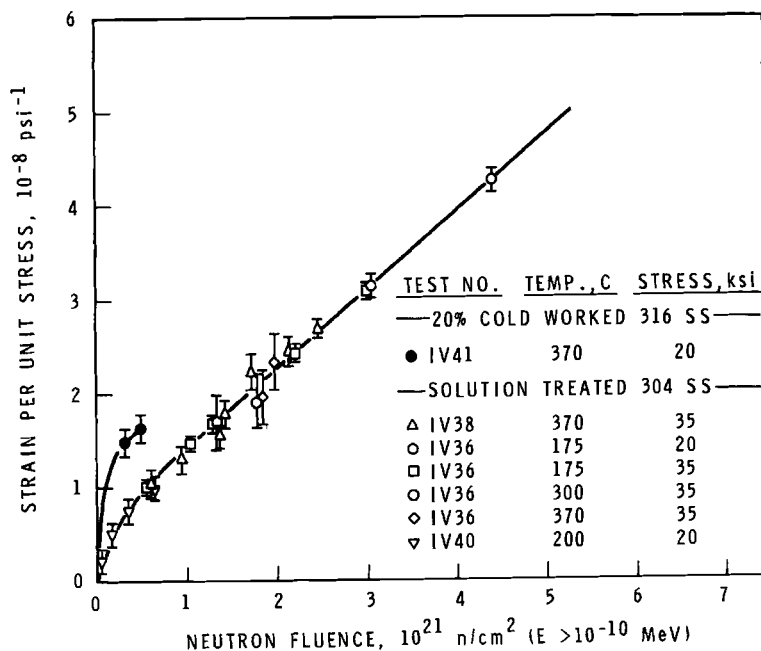


FIG. 7—In reactor creep data for austenitic stainless steels[15].

stresses for these experiments were sufficiently low so that strains were saturated after a few hours, and further deformation was not observed in the absence of neutron irradiation. When the experiment was placed into the reactor, creep strains were observed which appeared to follow a transient and steady-state behavior[15]. Information of this type is pertinent to the design and performance of stainless steel components in fast reactors. In some components, for example, irradiation-induced creep is expected to be beneficial by permitting relaxation of thermal stresses and stresses introduced by irradiation induced swelling.

Reactor Position

The selection of a reactor and specific location in it are important because of neutronic and gamma heat considerations. The neutron environment of one reactor is seldom duplicated in another due to variations in design, fuel, moderator and mode of operation. In fact, the neutron environment changes substantially within a single reactor not only as a function of axial and radial distance from the core center, but also because of the localized influence of control rods, fuel burnup, and composition of adjacent capsules, all of which may change even while an experiment is in progress.

One of the efforts in capsule design is to minimize neutron flux gradients. This is relatively easy for low flux irradiation facilities but difficult in certain positions in high flux experimental and power reactors[16]. At times, materials test specimens have to be miniturized to accommodate this limitation.

Another factor to which close attention has to be paid is the heat generated due to gamma radiation. Gamma rays react with the orbital electrons of a metal to produce heat, and in many cases the experimental capsule is designed to take advantage of this process for heating of the specimens. In certain cases the heating can be a detriment if it raises the temperature above the one desired for irradiation. Gamma heating of iron and uranium is illustrated in Fig. 8 as a function of radial distance from the center of the core of the experimental breeder reactor (EBR-II). Heat absorption is greater for uranium than iron, since it is a function of the atomic number of the elements. Furthermore, gamma heating decreases only slowly over the first five rows of the reactor core (out to a distance of 8.0–9.3 in. depending on specific position)[17]. The rapid drop off in gamma heat in Rows 6 and 7 makes temperature control much more uncertain.

An example of the use of gamma heat in the design of a noninstrumented specimen holder is shown in Fig. 9. The key to temperature control is the gas gap between the inner and outer containment capsules which provides a barrier to radial heat flow and allows temperatures greater than that of the reactor coolant to be obtained[18]. The sizing of

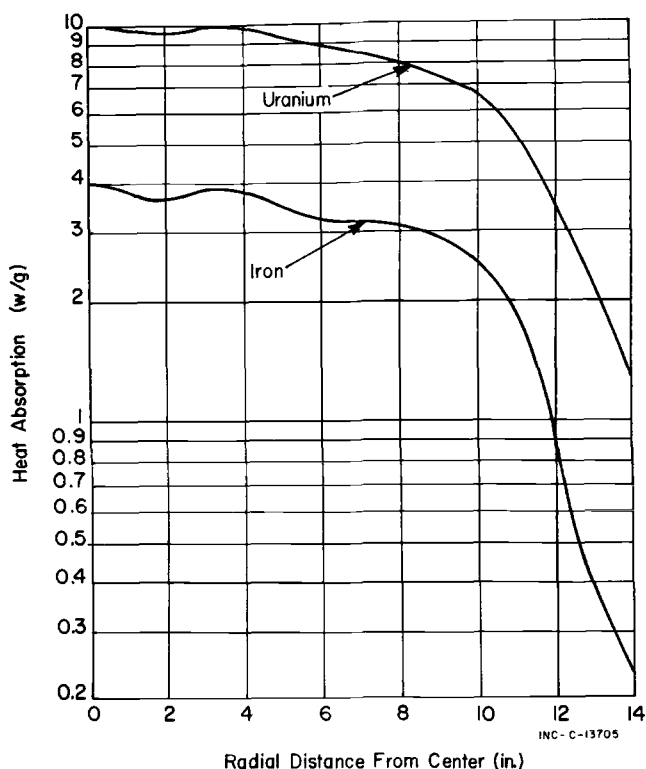


FIG. 8—Gamma heating at EBR-II midplane at a power of 50 MW[17].

the gap and the conductivity of gas used (helium, argon or mixtures of the two) provide some flexibility in the range of temperatures that can be obtained. Reactor position determines the amount of gamma heating of the specimen and the maximum temperature that can be obtained. For a gamma heating rate of 3.5 W/g, the calculated temperature of the

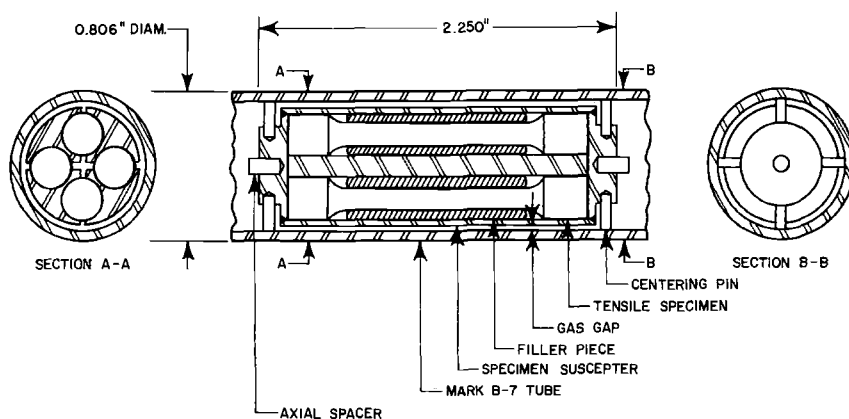


FIG. 9—Schematic of EBR-II specimen holder[18].

stainless steel specimens was 600 C compared to a maximum reactor coolant temperature of approximately 430 C on the outside of the outer containment tube[18].

Characterization of Reactor Environment

Neutron Dosimetry

In order to characterize the damage done by neutron irradiation, the conditions of exposure, specifically neutron flux, neutron spectrum, and neutron fluence have to be known within the limits of accuracy possible. The techniques used are primarily calculations based on fission spectra or else are measurements made using reactor beam ports supplemented by calculational techniques[19–22].

Direct measurements of neutron flux make use of the fact that neutrons can interact with a given target nucleus in different ways including (n, α) , (n, γ) and $(n, \text{fission})$ reactions. There is a specific cross section for each reaction, that is, the interaction taking place between a neutron and target nucleus is a fixed, reproducible value. Complete determination of cross sections is complex since they are a function of neutron energy, differ with the specific neutron reaction and vary even as a function of the isotopic content for a given element. The complete use of these reactions and the detectors used are reviewed in Ref 20.

A method for measuring reactor spectra by multiple foil activation has been devised recently[21,22]. This method can be used for individual experiments and is relatively fast and inexpensive, requiring no special facilities apart from a good radiochemical capability and the specific computer codes already developed. In this technique, a set of activation monitors of varying energy response is irradiated. The activities are considered as coefficients of the functional expansion of the spectrum in terms of the response curves. The mathematical problem is that of unfolding the spectrum from the coefficients. Such unfolding is performed with computer codes using energy dependent activation cross sections and a trial spectrum to calculate the expected activations for the materials comprising the monitor set, followed by comparing these with the actual activations and readjusting the spectrum by an iterative process[21].

The output of such a computer run after several iterations is illustrated in Fig. 10 where integral flux is plotted as a function of axial distance for neutrons of varying energies for the second row of EBR-II. Multiplication of such spectral data by the time at power, yields the fluence or total exposure.

Activation and fission monitors can be in the form of foil, wire, or pellets. A known mass of the material is placed in the neutron environment for a measured period of time followed by removal and counting.

EBR-II RUN 31F - RADIUS = 5.2CM
INTEGRAL FLUX ABOVE SPECIFIED ENERGIES

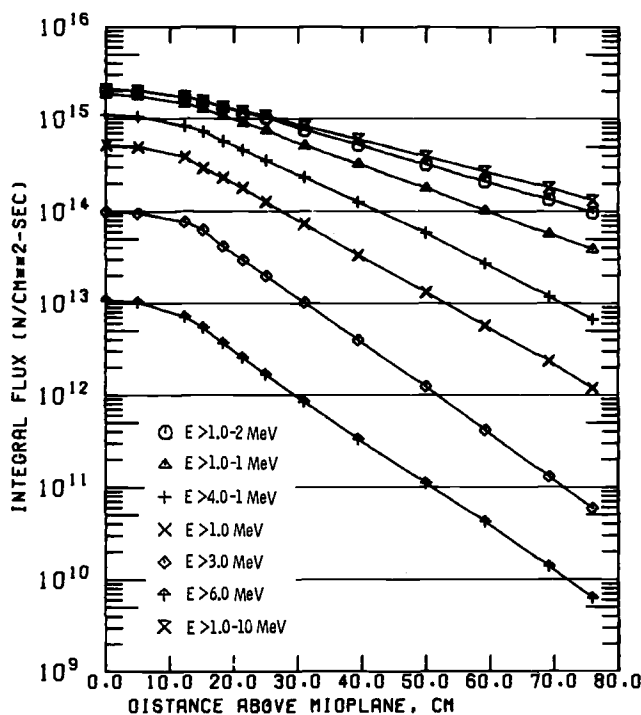


FIG. 10—Integral flux results of multiple foil activation for EBR-II row 2 position[23].

The type of dosimetry capsule used in EBR-II is shown in Fig. 11. The overall length of the stainless steel container is only 0.75 in. sufficient to hold the fourteen monitors used to determine the spectra at a specific location. In fact, the capsule is large enough to hold additional specimens if desired. Each monitor is kept apart from the others by its own containment, thus preventing cross contamination. Vanadium is used for encapsulating certain monitors, since its low cross section permits counting of radioactive species without removal or opening of the container. In addition to selecting monitors on the basis of sensitivity to the energy regions comprising the major portion of the spectrum, the following criteria should be met[21]:

1. The melting point of the monitor should be above the temperature anticipated for the environment.

2. The half-life of a product isotope should be long enough (on the order of days or months) to allow ample time for removal from the reactor and counting.

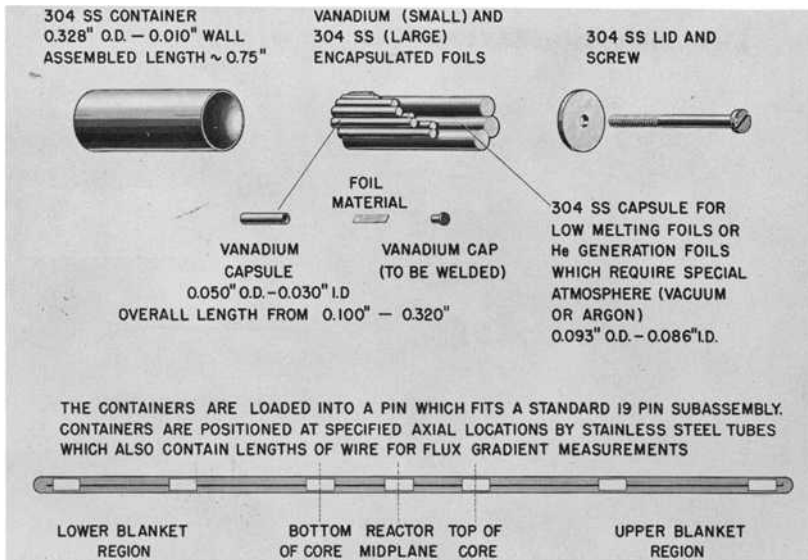


FIG. 11—Fast reactor dosimetry capsule[24].

3. The activation cross section for the material (and fission yield for fission foils) must be well defined.

4. High purity materials must be used in order to reduce analytical problems caused by the interference from the activation of contaminants. Flux perturbation and self shielding effects that could affect the accuracy of the measurements must be eliminated.

The response of the monitors selected for use in EBR-II is illustrated in Fig. 12. Each of the bars indicates the energy range that accounts for 90 percent of the product isotope activity at an in-core position in the reactor. The hatched area represents 90 percent of the neutrons in the reactor at that particular location. Activation of monitors falling outside of this range, such as titanium for instance, occurs by only a very small percentage of the neutrons, in this case those neutrons having energies from 6 to about 15 MeV. Repeating this experiment for a different part of the core changes the energy range which comprises 90 percent of the neutrons at the new location, and in turn changes the response of the monitors. The specific monitors selected can provide information on spectra in EBR-II ranging from the core center to positions well outside the core[23].

The limit of accuracy of neutron dosimetry required for structural irradiation is related to the uncertainty in mechanical property measurements. The multiple foil iterative method of neutron spectra determination has been shown to be accurate to within ± 10 to 30 percent. This is in the range where the total error due to dosimetry measurements is

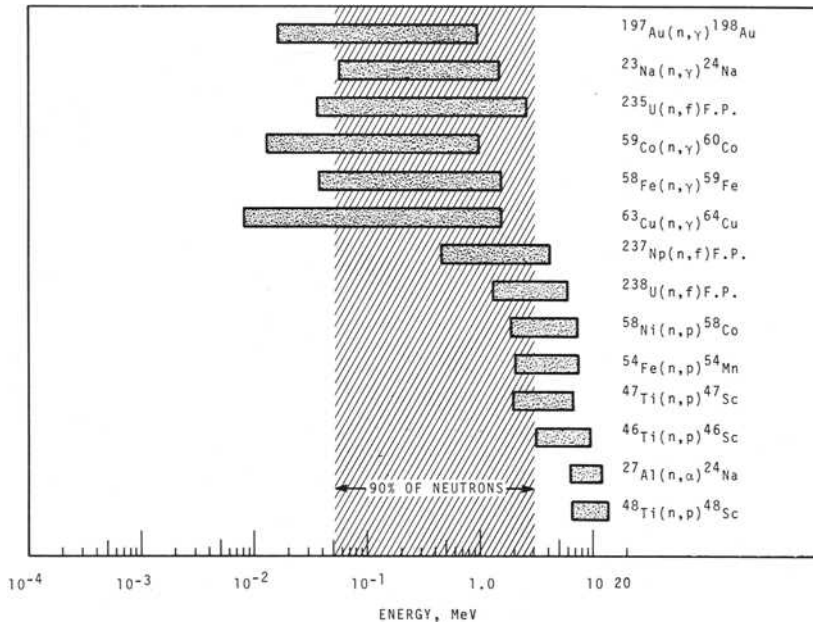


FIG. 12—90 percent response range of 14 dosimeters irradiated at core center in EBR-II[23].

no greater than that normally encountered in measuring the irradiation induced property changes.

Temperature Measurement

The importance of obtaining accurate temperature measurements during irradiation cannot be overemphasized. Since irradiation damage is a function of irradiation temperature, one needs to know the temperature quite accurately in order to predict the performance of specific materials or components. In spite of this need, the accuracy of temperature measurement in-reactor has not yet reached a very satisfactory level.

The most accurate and dependable way of measuring temperature is by way of thermocouples in instrumented capsules. An excellent review of the design of such thermocouples is given in Ref 25. The choice for most of the measurements up to 1000 C is Chromel-Alumel because of its good emf stability in the nuclear environment. In that temperature range, MgO has been found satisfactory as insulating material while Inconel is used as sheathing, except in cases where special compatibility requirements dictate another choice.

At temperatures greater than 1000 C, the choice of a suitable thermocouple material becomes more difficult. The noble metal thermocouples invariably contain rhodium which has a high transmutation rate with re-

sulting emf decalibration. Such refractory metal thermocouples as Nb-Mo, Nb-W, Ta-W, Ta-Mo and W-Mo are all unstable from an emf standpoint and have polarity reversals in the temperature region of interest, while the high rhenium containing Re-Mo and Re-W thermocouples are very unstable in a neutron environment due to rapid transmutation of rhenium. The best candidates for high temperature use are the tungsten-low rhenium combinations such as W3 percent Re-W25 percent Re.

An experiment was conducted recently in which a number of Chromel-Alumel thermocouples (Inconel sheathed, MgO insulated) and W3 percent Re-W25 percent Re (tantalum sheathed, BeO-insulated) were calibrated in-pile by continually monitoring the emf change for a fixed temperature[26]. High purity copper was selected as the fixed melting point element for the calibration cell. The experiment was performed in the Engineering Test Reactor (ETR) to an accumulated neutron exposure of 2.7×10^{21} n/cm² fast ($E > 0.18$ MeV) and 1.6×10^{21} n/cm² thermal. The duration of the test was 4470 h. Results are presented in Fig. 13 and show that the Chromel-Alumel thermocouples did not deviate significantly from their original calibration, whereas the W3 percent Re-W25 percent Re thermocouples exhibited a continuous deg-

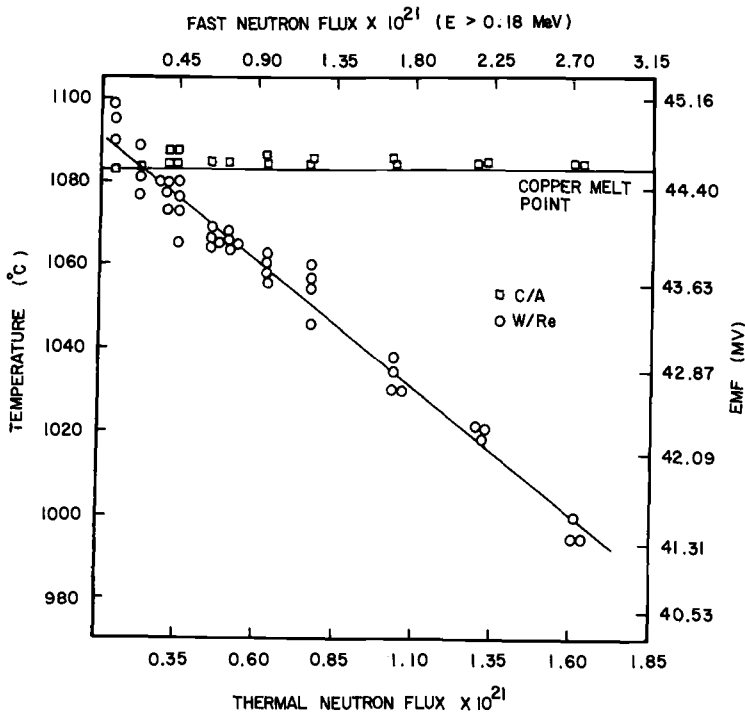


FIG. 13—Effects of neutron exposure on thermocouple stability[26].

radiation of emf output with increasing neutron exposure. The final decalibration was about 90 C for the particular test conditions.

The results of this experiment underline the importance of calibrating thermocouples in-pile. For the present, this still needs to be done on an individual basis since composition as well as neutron flux and spectrum probably have an effect. As more experience is accumulated and a better understanding is obtained, it may be possible to generate standard decalibration curves as a guide to radiation effects experimenters.

The principle means for measuring temperatures in-pile in noninstrumented experiments are silicon-carbide samples and fusible melt-wires. Silicon-carbide monitors work on the basis that fast neutrons induce dimensional changes which saturate as a function of fluence at a given temperature. Upon postirradiation annealing, very little shrinkage occurs until the annealing temperature exceeds the irradiation temperature, at which point a linear change in dimension occurs with increasing temperature. The break in the curve thus reflects the irradiation temperature. Temperature measurements of this kind are illustrated in Fig. 14. Three of the five "break points" indicated a maximum temperature ranging from 710–750 C, and melt wires within the same experiment were in agreement that a maximum temperature of 750 C had been reached[27].

The principle of melt-wire operation is conceptually simple inasmuch as it relies on a series of specimens of known and different melting points covering a specific temperature range. Post irradiation examination or sometimes nondestructive surveillance examination, then determine which wires have been molten, indicative of the highest temperature attained. In practice, however, several difficulties may be encountered, including the availability of a series of wires that exhibit a sharp melting point in the temperature interval of interest, and the exact determination of which wires have been molten during the experiment. Good characterization of melt-wire compositions is required. In general, elements and eutectic compositions have the sharpest melting points, but solid solutions are frequently employed in order to increase the number of melt-wires of different melting points. To increase the sensitivity of determining whether a wire has been molten, wires are often machined with a pointed end or a groove in the center. The latter technique has been used successfully to inspect wires radiographically where a radiation hazard was involved, thus making visual inspection unnecessary[28]. A series of melt-wires covering the 100–1098 C temperature range is illustrated in Fig. 15. The published melting point of each composition is marked by an X, whereas the bar width indicates the temperature spread between 0 percent and 100 percent melting in calibration tests conducted on twelve wires of each composition. The encapsulation of the wires in quartz under vacuum reduced the scatter in melting points observed, since it reduced reaction with the outgassing products[28]. On the other

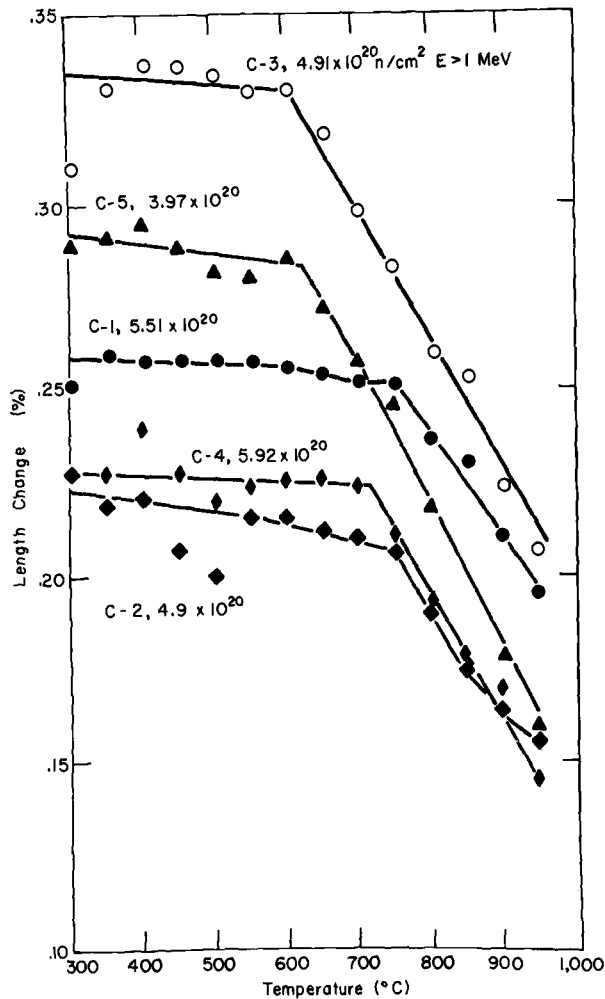


FIG. 14—Annealing response of irradiated SiC temperature monitors[27].

hand, accurate temperature measurement becomes more difficult when the wires are encapsulated.

The disadvantages of both silicon-carbide and melt-wires are that neither measurement gives any information on the length of time the irradiated specimens were at temperature. A short time-temperature excursion will affect the measurements and may lead to erroneous conclusions. In addition, the post irradiation measurements are tedious and time consuming. On the plus side, however, it must be considered that temperature calculations based on heat transfer considerations can be off by as much as ± 100 C. A well conducted experiment in temperature measurement of the type described can reduce this uncertainty to ± 10 – 15 C.

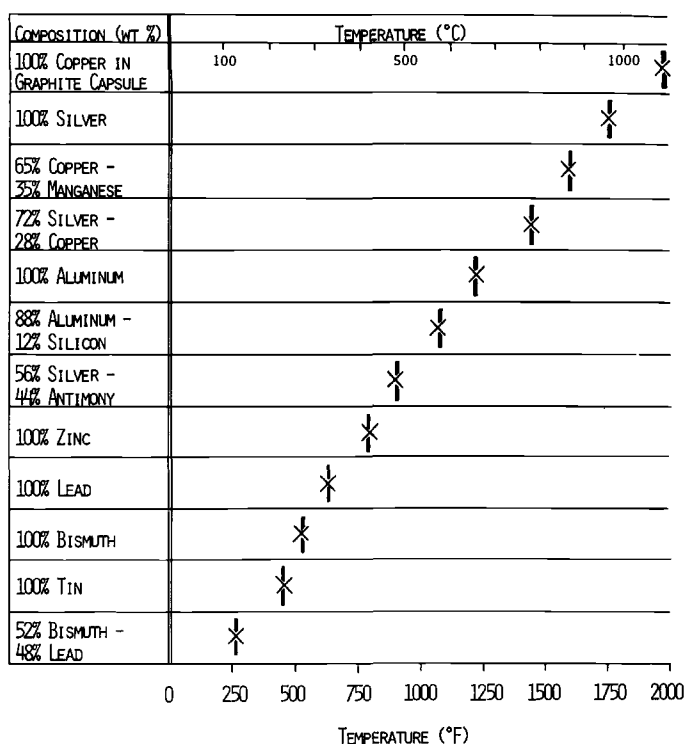


FIG. 15—Characterization of melt wires with melting points below 1098 C[28].

Examples of Predictive Testing

Predictive testing of structures and components is sometimes difficult to achieve in-reactor, because of severe space limitations and environmental conditions. A pressure vessel, for instance, cannot be irradiated as a structural entity because of its size, and specimens of the same material, often from the same heat used to construct the plates for the vessel, are irradiated and tested instead. This has given rise to extreme conservatism in the design of nuclear-reactor vessels because of the importance of the vessel in forming the pressure boundary between the nuclear environment and the outside.

In the case of liquid metal cooled reactors, materials can be tested in-reactor and separately within the liquid metal outside of the reactor. Yet testing in-reactor under the exact environmental conditions is extremely important. Not only does this apply to materials in direct contact with the coolant, but the compatibility of coolant and contained materials such as the fuel or moderator must be investigated in case of a failure of the containment and direct interaction between the two. This is illustrated in the Hallam example below.

It must be kept in mind that often one reactor is used for predictive

testing of components to be used in another reactor just being built, with possible differences in flux, spectrum, temperature profile, and coolant. Data have to be obtained over the widest range of experimental conditions possible in the test reactor to minimize extrapolation.

After all materials, components and structures have been tested to the best techniques available, reactors are started up and run in the most conservative fashion. A series of surveillance tests is used to assure that all components operate in the manner predicted. Fuel assemblies are removed from the reactor after relatively short intervals to examine pin behavior and to check for pin-to-pin interaction. Ducts, control rods and other critical components are inspected. Specimens of pressure vessel steel, located close to the wall of the vessel are removed periodically and tested destructively, to follow the changes in properties occurring in the vessel wall. The fuel behavior in EBR-II discussed below, is a good example of the value of surveillance testing and the care that must be taken in evaluating even relatively minor changes in an established process.

Improved Radiation Resistance by Control of Residual Elements

Pressure vessels for the light-water-cooled type of reactors in the United States are fabricated from carbon or low-alloy steels. Neutron irradiation increases the hardness and yield strength of such steel while decreasing ductility, especially notch ductility. Predictive testing for these steels has been carried out for a long period of time and is particularly fascinating because of the many variables involved. These include the location of specimens within the plate (surface, quarter thickness, center), direction of specimens (transverse, longitudinal), type of specimen (Charpy-V, drop weight test, and others), heat treatment, irradiation temperature, flux, fluence, and spectrum effects. Excellent reviews on the results of such studies are available[29,30].

For the past five years special emphasis has been placed on studying the effect of residual elements on notch ductility after irradiation. Like many advancements in the state-of-the-art, this one had its origin in an unrelated investigation[31]. A number of A212-B heats containing uranium additions were being investigated. The findings indicated superior properties after irradiation which, however, were not a function of uranium content. In looking for other effects it was discovered that the heats were of higher purity than usually encountered, and further laboratory studies revealed a strong influence of residual element content on radiation embrittlement[32]. In particular, copper and phosphorus were identified as having the greatest potency in affecting embrittlement. In order to translate the findings made on laboratory heats to production quantities, a scale-up demonstration test was conducted on A533-B steel involving 6-in.-thick plates from a 30-ton melt[33]. The effect of copper content (at a constant phosphorus level of 0.008 percent) is demon-

strated in Fig. 16. Comparison at a Cv energy of 30 ft·lb shows an increase in ductile-to-brittle transition temperature of 22 C (40 F) for the 0.03 percent copper, compared to an increase of 78 C (140 F) for the 0.13 percent copper content. Closed data points in the figure refer to the longitudinal orientation, while half-filled points refer to the transverse direction, indicating the same effect in either measurement.

The effect of copper on A543 weldments is illustrated in Fig. 17. The increase in Cv transition temperature at 30 ft·lb was less than 11 C (20 F) for copper contents of 0.05 percent or less, while being in excess of 111 C (200 F) for copper contents ranging from 0.21 to 0.28 percent[34].

At the present time, the above findings are under evaluation by ASTM Committee A-1, with a view towards supplementing ASTM specifications by placing composition limits on several elements including copper, phosphorus, sulfur and vanadium. Predictive testing in this case lead to

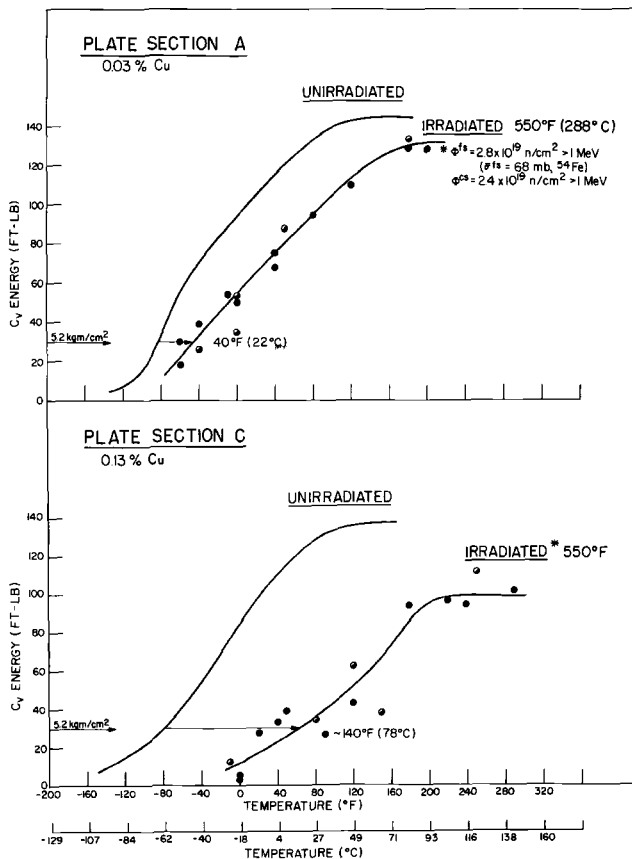


FIG. 16—Effect of copper on notch ductility of A533-B steel[33].

EXPERIMENTAL WELD PERFORMANCE COMPARISON

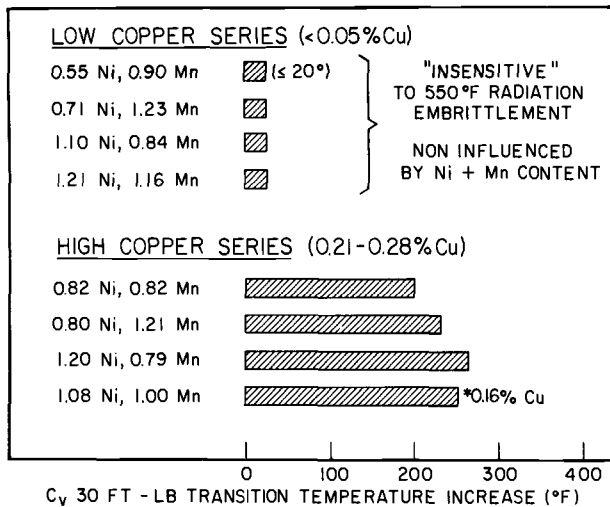
(IRRADIATED 550°F (288°C), 2.8×10^{19} n/cm² > 1 MeV)

FIG. 17—Effect of copper on notch ductility of A543 weldments[34].

improvements in alloy composition, resulting in greater safety factors for water cooled reactor pressure vessels.

Failure of HNPf Moderator Elements

The Hallam Nuclear Power Facility (HNPf) was a thermal, sodium cooled reactor containing graphite as a moderating material to lower the average neutron energy[35]. The purpose of the reactor, in addition to being a power source, was to provide a simpler core—compared to a fast sodium cooled reactor without moderator—around which large sodium components and piping systems could be developed and tested. From July 1963 through September 1964, HNPf produced close to 200 million kwh of electricity[35].

In February of 1964, losses of reactivity associated with temperature depressions in the core were observed and diagnosed as failures of individual moderator elements. Prior to plant shutdown in September 1964, seven out of a total of 141 moderator elements had failed and were logged with sodium[35,36]. As a result of the sodium soaking, the graphite swelled more than expected, leading to a jamming together of some moderator elements, fuel elements, control rods, dummy elements, and the source element in the core of the reactor. In order to remove the moderator elements the core had to be unloaded and auxiliary equipment had to be designed and built, operations that took about 10 months to perform.

Complete failure analysis including properties of the materials involved, theoretical stress analysis, experimental stress analysis, design

and fabrication reviews, microscopy and examination of the behavior of graphite in sodium were carried out on the failed elements and on available spares[36,37]. The basic failure consisted of stress-rupture cracks in the 304 stainless steel sheath used for encapsulation of the graphite blocks. This was followed by expansion of the sodium logged graphite to an unexpected degree, leading to the jamming of reactor components.

Stress analysis identified the cause of the high stresses leading to stress-rupture failures to be lack of support of the sheath by the ribs extending from the head of the element. While the geometry of the elements and other factors all played complicating roles in the analysis failure essentially occurred by an off design failure mode due to stress concentration at a particular support point of the sheath, whereas a stress distribution over a number of support points had been postulated.

The HNPF moderator element design was intended to accommodate graphite expansion, in the event the latter became soaked with sodium. Such graphite expansion was predicted to be 1 percent on the basis of tests on unirradiated graphite. The dimensions of graphite logs were measured on the failed elements and showed up to 6 percent increase in width. Empirical observations led to the conclusion that graphite swelling in sodium was radically different at a neutron dosage greater than 3×10^{20} n/cm² than at lesser exposure. A rough estimate of swelling could be made from the data and at neutron exposures $>3 \times 10^{20}$ n/cm² ($E > 0.18$ MeV), the HNPF moderator graphite swelled 2 percent in <10 h, whereas at exposures $<3 \times 10^{20}$ ($E > 0.18$ MeV) it took 4000 h to swell the same amount. In several instances the bottom or top head of the moderator element or both was separated from the sheath due to graphite swelling.

The failure of HNPF moderator cans is a good example of interaction or synergistic effects that must be evaluated in predicting component performance. Although the primary failure was perhaps principally one of design and not related to radiation, the secondary effects—swelling of irradiated graphite in sodium—caused the more detrimental effect of jamming of core components, making it difficult to remove and replace the failed elements in the reactor.

Shortening of EBR-II Fuel Elements

The fuel for the EBR-II Reactor is a uranium—5 w/o fissium alloy.² The fuel element consists of a pin produced by injection casting which is sodium-bonded to a Type 304L stainless steel jacket. Until late 1968, all elements were produced by Argonne National Laboratory (ANL) by

² Fissium (F_s) is an aggregate composed of molybdenum, ruthenium, rhodium, palladium, zirconium, and niobium. The composition is the equilibrium alloy of the major fission products left during pyrometallurgical reprocessing of the fuel.

either a "hot line" process consisting of reprocessing spent fuel into new pins, or by a "cold line" process which simulates the composition from virgin materials. Elements made by either process exhibited isotropic growth during irradiation, and tens of thousands of pins were successfully run through the reactor.

As part of a program to qualify a commercial fuel supplier to fabricate driver fuel elements, vendor produced fuel was inserted into EBR-II in late 1968. A surveillance program was initiated under which fuel was removed periodically from the reactor and inspected for any anomalous behavior. In the spring of 1969, initial irradiation results showed that the vendor produced fuel had shortened during irradiation. The change in length was 2.3 cm for pins initially 34.2 cm long with accompanying radial growth varying from zero to 8 percent along the length of the pin after 0.6 atom percent burnup[38].

A number of mechanisms were postulated for the pin shortening, among them preferred orientation induced during the fuel fabrication process. Such preferred orientation was indeed shown to be the case[39], and it remained to find the step during processing responsible for this. A careful comparison between the ANL and vendor fuel element fabrication processes revealed only one significant difference—that of the sodium bonding step. The ANL elements were heated to 500 C and then impacted many times to eliminate bubbles in the sodium between the pin and the cladding to achieve a good bond. The vendor elements were centrifugally bonded and a centrifugal force was maintained both during heat up and during the actual bonding operation at 500 C.

In order to understand the importance of the applied force during bonding, one needs to examine the phase relationships in the uranium-fissium alloy. The as-cast microstructure can retain between 65 to 90 percent of the elevated temperature γ phase depending on cooling rate. During bonding at 500 C the γ phase transforms to the equilibrium phases $\alpha + \text{U}_2\text{Ru}$ with the matrix being α . In the ANL bonding process, the retained γ transformed to α uranium before impacting. In the vendor process, the transformation occurred under an applied stress resulting in a preferred orientation of the α matrix phase. This, in turn, resulted in the anisotropic growth observed during irradiation[40].

The foregoing is a good example of how relatively minor changes in processing or fabrication can affect irradiation performance. While preferred orientation had been observed for uranium alloys, it had been associated with extruded or rolled fuel but not with cast material. Preferred orientation produced by an externally applied stress during phase transformation also had not been reported previously for uranium[40]. EBR-II fuel shortening also is a good example of the value of surveillance testing. Careful surveillance of any material placed in a reactor for the first time is necessary to insure that it behaves according to prediction.

Void Formation in a Fast Neutron Environment

The formation of voids in metals during fast reactor irradiation was first reported in 1967[41]. The voids result in swelling of components and the changes in dimensions must be taken account of in the design of fast reactors. An illustration of voids in Type 304 stainless steel is given in Fig. 18. The voids are on the order of 300 Å in size, in some cases are faceted along crystallographic planes and in general are absent in the region of the grain boundary[42]. These features have been observed



FIG. 18—Voids in irradiated type 304 stainless steel[42] (Distance between marks equals one micrometer.).

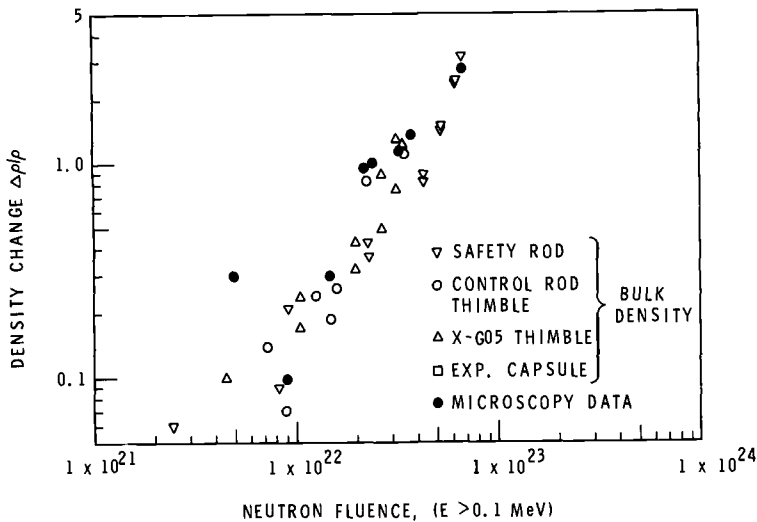


FIG. 19—Swelling as a function of fluence for several EBR-II components[42].

many times by different investigators, and void sizes varying from less than 100 Å to more than 1000 Å have been observed. Figure 19 presents data for density or volume change as a function of neutron exposure. The data were obtained on a number of components removed from EBR-II and show swelling up to 4 volume percent at an exposure of 7×10^{22} n/cm² ($E > 0.1$ MeV). Transmission electron microscopy and bulk density measurements both confirm these observations. The effects of structure and composition are illustrated in Fig. 20. Commercially pure nickel exhibits the greatest amount of swelling as a function of temperature. Cold-working Ni-200 by 50 percent reduces the swelling markedly, at least up to that temperature where the metal recrystallizes and its properties become those of the annealed material. Annealed as

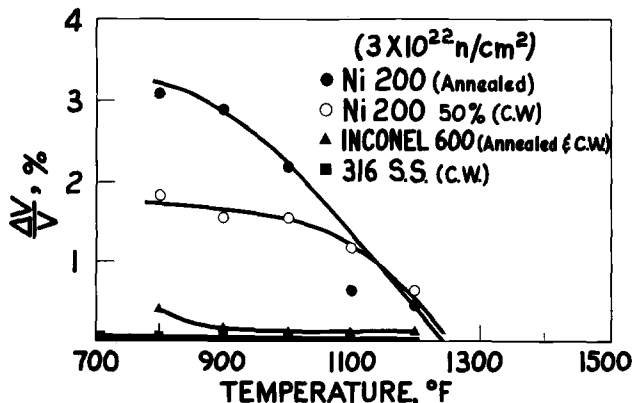


FIG. 20—Swelling as a function of irradiation temperature for several alloys[43].

well as cold-worked Inconel and Type 316 stainless steel by comparison, exhibit no swelling at these temperatures and for the particular neutron exposure indicated[43]. However, at higher exposures, these alloys also are subject to swelling. The importance of swelling from a practical point of view is threefold. First is the potential closing of coolant passages resulting in changes in the hydraulic characteristics of the core. Second are the stresses introduced within a component, the various parts of which operate in a different flux or temperature environment and therefore swell at different rates. Third is the possible interaction of a number of components which swell at unequal rates because of differences due to composition, temperature, or flux.

It follows from the discussion above that predictive testing plays an important role in establishing the performance limits due to swelling of materials and components in a fast flux. One must now consider the variables and interactions between them that must be taken account of in a test program. To begin with, swelling has been observed in almost all metals and alloys of interest in sodium cooled fast reactors including nickel and nickel-base alloys, iron and iron-base alloys including stainless steels, vanadium, molybdenum, tantalum, and niobium. The usual temperature range is $0.3T_m < T < 0.5T_m$. Swelling generally is greater in pure metals than alloys, goes through a maximum as a function of temperature, increases with increasing fluence, is a function of cold work or initial dislocation density and may be increased by an applied stress. The latter point is still in doubt and evidence exists both for and against it. Void diameter has been found to increase with increasing irradiation temperature while void number density decreases.

A number of theories have been advanced to explain these effects. It is generally agreed that void formation is due to a strain field biased attraction of interstitial atoms to dislocations resulting in a net excess of vacancies. Void growth then occurs by the diffusion of vacancies to the void[44-48]. Although there is fair agreement among investigators on the mechanisms responsible for void formation and growth, mathematical formulations need to be refined for accurate predictions of swelling under varying conditions.

In a discussion of predictive testing to sort out the effects of these variables and to provide design data, it must be kept in mind that it takes on the order of two years in EBR-II to arrive at an exposure deemed sufficient to allow separation of variables, and a minimum of four years to reach the end-of-life design exposure for the fast test reactor (FTR) currently being designed. A great deal of ingenuity has been applied in the design of experiments now in progress to obtain the necessary confirmatory information for FTR. Every effort has been made to cover all variables so that accurate swelling numbers can be assigned to any combination of conditions. In this particular example of predictive testing, the data

obtained will be combined with careful surveillance testing at the start-up of FTR, to assure that the predictions made on the basis of available data are indeed correct.

Current Needs and Goals

There is a great deal that remains to be done in predictive testing of irradiation effects. To the extent that test reactor space is limited, one must continue to be selective in the planning and execution of experiments. Testing space for materials and components for the liquid metal fast breeder reactor, the Atomic Energy Commission's highest priority civilian power program, remains severely limited since only the EBR-II is available for anywhere near prototypical testing. The startup of the FTR will, for the first time, provide the high flux necessary for prototypical testing of materials used in commercial fast reactor plants. The high flux—on the order of 7×10^{15} n/cm² s or about three times that attainable in EBR-II—is necessary to obtain end-of-life exposure for structural materials within a reasonable period of time and to obtain the correct fluence to burnup ratios for fuel pin cladding. In addition, the FTR contains a number of closed sodium loops, completely isolated from the rest of the system. Such loops will be used to test fuel to failure and beyond, as well as to simulate any special environment desired—all without interfering with the balance of reactor operation or without contamination of the primary coolant.

Within the current programs, a great deal of emphasis is being placed on gaining a better understanding and mathematical description of reactor environments, specifically neutron flux, spectrum and temperature. Refinements in measuring these parameters are needed in order to more accurately predict radiation induced changes in properties. For this reason, cross sections of activation and fission monitors are now being redetermined or measured accurately for the first time with the advantage of new and improved techniques. Gamma heat measurements, likewise, are being repeated and refined for test reactors such as the EBR-II.

It has been recognized that correlation of mechanical properties simply on the basis of neutron fluence of energies greater than a certain cut off point such as 1 MeV is not sufficiently accurate for critical applications. A method has been developed recently for generating damage functions for the correlation and prediction of neutron-induced property changes[49]. With this procedure it will be possible to determine the neutron spectral dependence of irradiation effects data obtained in test reactors, with subsequent extrapolation and prediction of component performance to power reactors. The method has been found to work well when applied to the ductile-to-brittle transition temperature increase data obtained in different spectra[50].

In regard to the influence of neutron irradiation on elevated temperature

mechanical properties, more accurate cross sections for helium reactions are now being obtained. In addition, transmutation effects resulting from the production of metallic isotopes need to be better understood along with radiation enhanced nucleation of precipitates, migration kinetics of displaced atoms, and the interaction of radiation induced defects with dislocations.

During the past twenty-five years, radiation effects testing has progressed to the point of fulfilling the needs of a new and growing technology. While further improvements in testing techniques will undoubtedly be made, primary advancements in predicting neutronic effects most likely will come from a better understanding of the specific contribution to damage by neutrons of varying energies.

Acknowledgments

We wish to acknowledge that the material in this paper is based on the excellent and dedicated work of many people in several Atomic Energy Commission laboratories, and many other organizations both here and abroad.

We also wish to express our appreciation to the individuals who supplied the illustrations used in the paper. Each illustration is referenced as to its original source.

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Techniques for Smooth Specimen Simulation of the Fatigue Behavior of Notched Members

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ABSTRACT: The initiation of fatigue cracks in a notched member is determined by the local cyclic stress and strain at the notch root. A relationship, proposed by Neuber, is used to relate the notch root stress-strain response to the nominal stress in the notched member. Further, several methods based on Neuber's rule allow a smooth cylindrical specimen to be used to simulate the stress-strain behavior of the notch root metal. The fatigue life of the smooth specimen is shown to furnish a satisfactory estimate of the fatigue life of the notched member, and the stress-strain plot of the smooth specimen is useful in understanding the local effects on the notch root metal of various nominal load histories. Results are reported for notched specimens subjected to variable load histories and for the corresponding smooth specimen simulations for 2024-T4 and 7075-T6 aluminum, aircraft quality 4340 steel, and Ti-8Al-1Mo-1V.

KEY WORDS: predictions, stresses, fatigue (materials), notch sensitivity, loads (forces), cyclic loads, crack initiation, mechanical tests, softening, hardening (materials), fatigue life

To predict the fatigue crack initiation life of a notched member, one must know the local stress-strain history at the notch root and the fatigue resistance of the notch root metal.

Since the introduction in 1961[1] of Neuber's rule, which equates the theoretical stress concentration factor to the geometric mean of the stress concentration factor and the strain concentration factor, several investigators[2–6] have shown how this simple concept can be extended to relate the stress-strain behavior of the metal at a notch root to the load history on a notched specimen. This type of notch root analysis suggests that a smooth specimen, fatigue tested under a control condition satisfying Neuber's rule (called Neuber control) will respond with a stress-strain history similar to the metal at a notch root. Further, the fatigue life of

¹ Research assistant and professor, respectively, Department of Theoretical and Applied Mechanics, University of Illinois, Urbana, Ill. 61801.

the smooth specimen may be used to estimate the fatigue crack initiation life of the notched specimen being simulated[4,6].

Smooth specimen simulations transfer the complexity of predicting failure from the testing of the actual notched part under its load history to the testing of a smooth specimen under Neuber control. Ideally, Neuber control reduces the fatigue testing of complex parts to the testing of smooth, identical specimens.

This paper explores the feasibility, accuracy, and practical difficulties encountered in attempting to perform fatigue tests on smooth specimens under Neuber control and to evaluate the correlation of such tests with actual notch fatigue behavior.

Neuber's Rule

The basis for simulating the fatigue behavior of notched members, using smooth specimens is the equation presented by Neuber[1]:

$$K_t = (K_\sigma K_\epsilon)^{1/2}$$

where

K_t = theoretical stress concentration factor,

K_σ = stress concentration factor, and

K_ϵ = strain concentration factor

In fatigue applications, the fatigue notch factor K_f , is used instead of K_t . The value of K_f is usually determined from long life notched and unnotched fatigue data, and has been shown to be relatively independent of fatigue life[4,6].

For cyclic loading Neuber's equation may be written in the following form using the Δ symbol to indicate ranges:

$$K_\sigma = \frac{\Delta\sigma}{\Delta S}$$

$$K_f = \left(\frac{\Delta\sigma}{\Delta S} \frac{\Delta\epsilon}{\Delta e} \right)^{1/2}$$

$$K_\epsilon = \frac{\Delta\epsilon}{\Delta e}$$

where

$\Delta\sigma$ = stress range at notch root,

ΔS = nominal stress range,

$\Delta\epsilon$ = strain range at notch root,

Δe = nominal strain range.

The above equation may be factored giving

$$K_f (\Delta S \Delta e)^{1/2} = (\Delta\sigma \Delta\epsilon)^{1/2}$$

All terms on the left side are determinable for each reversal (half-cycle)

from the load history and cyclic stress-strain curves, and all the terms on the right side represent the local stress-strain behavior of the material at the notch root.

In most engineering applications, the metal away from the notch root remains elastic, so that Hooke's law relates the nominal stress and strain. Therefore $\Delta S = \Delta \epsilon E$, and

$$K_f \Delta S = (\Delta \sigma \Delta \epsilon E)^{1/2}$$

or

$$\Delta \sigma \Delta \epsilon = \frac{(K_f \Delta S)^2}{E}$$

The right side of the above equation is a determinable constant for each half-cycle. The equation is of the form, $xy = C$, which is a rectangular hyperbola.

For simulations of the metal at the notch root, the smooth specimen must be strained until the product of the stress and the strain equals the constant, C (where $C = \frac{K_f \Delta S)^2}{E}$).

Since all the quantities considered are the changes per reversal, the point of intersection of the hyperbola and the stress-strain plot must be used as the new axis for the next reversal (Fig. 1).

In summary, Neuber control is defined in this paper as the sequential application of the following control condition. These steps are repeated until the smooth specimen fractures:

1. Determine $\frac{(K_f \Delta S_1)^2}{E} = C_1$
2. Draw a hyperbola so that $\Delta \sigma \Delta \epsilon = C_1$
3. Stress (or strain) the material until its product of stress range times strain range equals C_1
4. Use the point of intersection as the new axis
5. Determine $\frac{(K_f \Delta S_2)^2}{E} = C_2$
6. Draw C_2 on the new set of axes
7. Repeat steps three through six for each additional reversal.

Figure 1 illustrates the steps involved in practical applications of the method.

Methods of Enforcing Neuber Control on a Smooth Specimen

Three methods of smooth specimen simulation control have been developed. For simple load histories manual control may be practical. The combination of a digital computer and a mechanical test system makes it possible to simulate more complex load histories. Finally, an analog approximation to Neuber control has been developed which will be called "sliding line control."

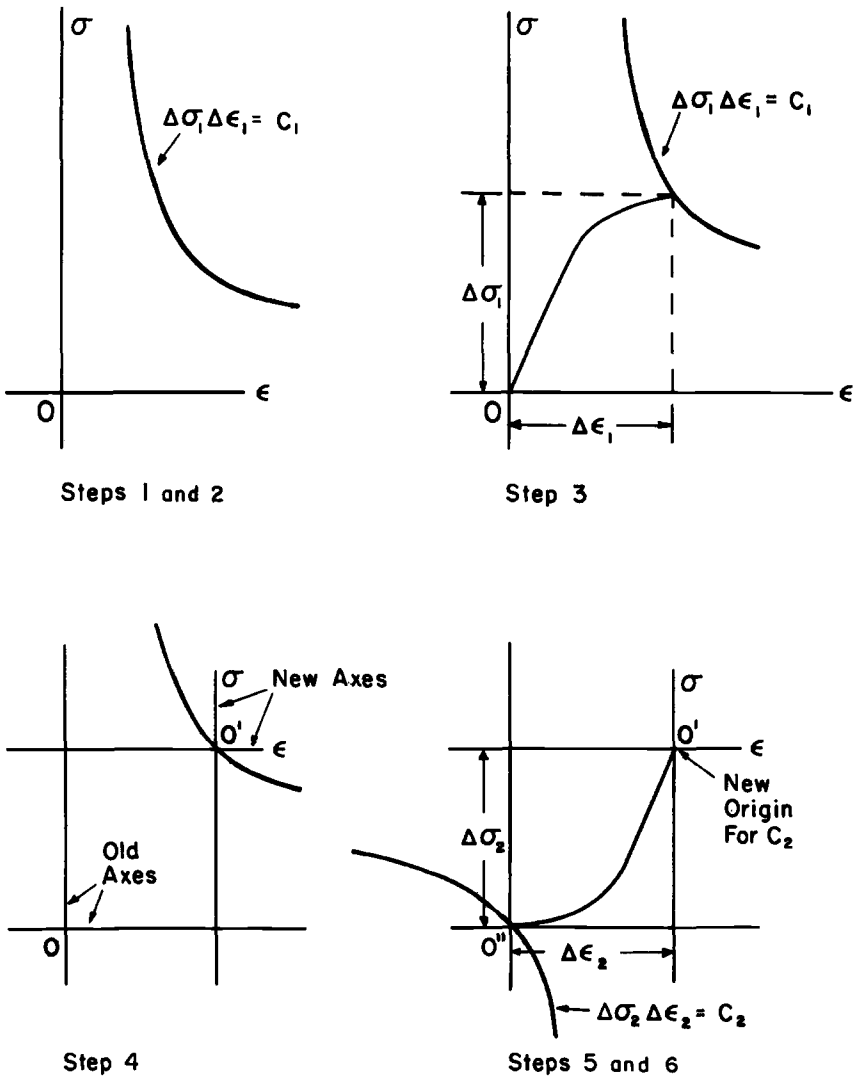


FIG. 1—Steps of Neuber control.

Manual Control

For simple tests the control hyperbola may be drawn onto a stress-strain plot. The specimen is loaded until the stress-strain curve reaches the hyperbola. Straining must be stopped at each reversal point and the next hyperbola drawn. Obviously, this crude method is practical for only a few reversals. Some other type of control, for example, strain control, must be employed after a few reversals to prevent the accumulation of error.

The errors in satisfying the hyperbolic relationship may be additive

under Neuber control, since the position of the new hyperbola depends directly on the intersection point with the last hyperbola. A small error per reversal can result in considerable drift of the stress-strain axis after a large number of reversals.

Despite the limitations on manual control it has been successfully employed by Wetzell[6] for the simulation of zero to maximum loading and by Stadnick[5] to simulate the influence of an initial large overload on the fatigue life of notched members. For more complicated load histories, automated methods of enforcing Neuber control must be used.

Analog Computer Control

An initial attempt was made to use a small analog computer in the control circuit of an electro-hydraulic materials test system to achieve Neuber control. However, experience to date has been disappointing.

The value of $(K_f \Delta S)^2/E$ was computed electrically and compared with the product, $\Delta\sigma \Delta\epsilon$ for each reversal. The difference in products was used as an error signal to control the movement of the hydraulic ram. Accurately rezeroing of the σ - ϵ axis, which is required each time the control hyperbola is encountered, proved to be a major source of error. The use of capacitors for storing the values of stress and strain at the starting point for each reversal for the calculation of the product, $(\Delta\sigma\Delta\epsilon)$, proved unsatisfactory.

Errors in rezeroing, operational amplifier drift in critical components, and poor resolution due to electrical noise all contributed to limit the usefulness of the analog control circuit. However, the effort did lead to the development of a useful approximation of Neuber control called sliding line control, which is discussed in a later section and the Appendix.

Digital Computer Control

Under digital computer control, the movement of the hydraulic ram is controlled by computer output[7]. The computer analyzes the stress and strain outputs of the closed loop axial fatigue test system and determines the direction of the next increment of ram displacement necessary to satisfy Neuber's relationship.

Primary advantages of digital control are accuracy and versatility. Since all subtractions and multiplications are carried out in digits within the computer, the resolution of the system is limited only by the input load and strain signals from the test system. The performance of digital computer control can be described by the following characteristics.

Accuracy—The load and strain voltages were converted to digital representation in steps of 0.05 V. Thus, the accuracy of determining the stress and strain at the point of each reversal is a function of the size of the Neuber hyperbola (see Fig. 2). Normal full scale voltage levels for

Error due to incremental
step conversion of stress
and strain voltages to
digital numbers — does not
contribute to drift

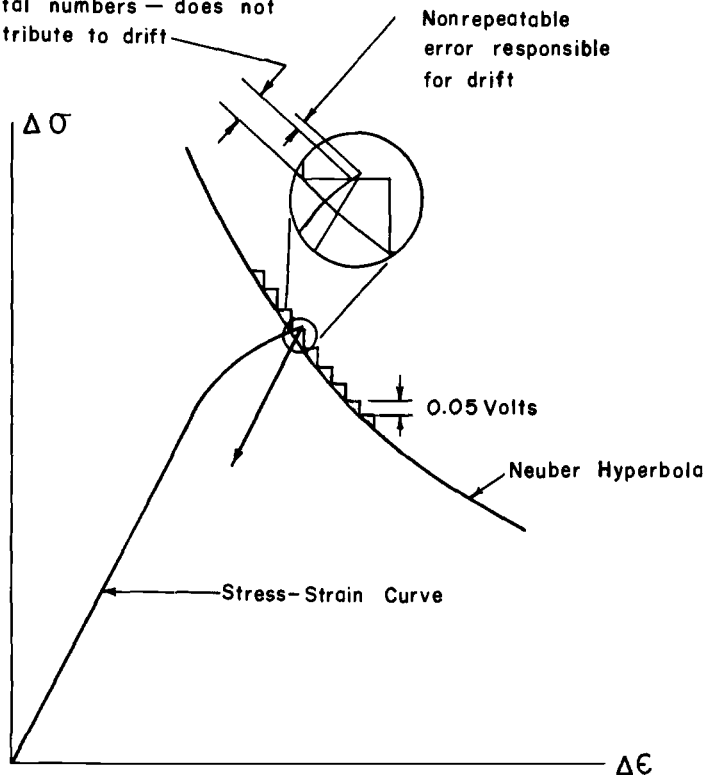


FIG. 2—Accuracy of digital computer control.

the stress and strain signals are ± 10 V, so that the maximum value of their product is 400 and the inaccuracy in control is trivial. However, small values of stress and strain, giving products in the range of about five, would result in a control error of approximately 1 percent.

Even though the control error at small stress and strain values for a single reversal may be considerable, the repeatability of computer control prevents the accumulation of significant errors, so that "drift" is not a serious problem. For example, the drift of the system was found to be less than 3 percent of the strain range in a test of an aluminum specimen which was subjected to 500 000 reversals under Neuber control at a small elastic strain level. For tests in the 5000 to 100 000 reversal range, this is an acceptably small drift.

Digital computer Neuber control was found to be very sensitive to induced voltage spikes in the stress or strain signals or both. Transients of very short duration, but of several volts in magnitude were interpreted by the computer as a load or strain signal. The transient stress-strain

product triggered a reversal even though the product of the increment in stress and strain had not yet equalled the value of the hyperbola. The net result of this early reversal was the same as if the zero of the load history on the simulated notched specimen had been shifted. Transients were filtered as much as possible without undue damping of the stress and strain signals, and are not a major problem with proper filtering.

Versatility—Since the digital computer can be programmed in advance, it can perform many additional tasks while simulating on a smooth specimen the fatigue behavior of a notched member. The following description of the program used illustrates the versatility of the digital system.

1. The program input included a synthesized load history of the notched specimen being simulated. This load history was divided into two blocks of 24 reversals each. These two blocks were programmed to run in a sequence, so that the first block was applied once, followed by repetition of the other block until failure. Single occurrences of initial overloads were simulated in this way.

2. The strain rate of the test could be changed at any time. It was possible to perform the simulation at one strain rate, and record significant hysteresis loops at a slower rate. The maximum ram displacement rate under Neuber control allowed maximum cyclic frequencies between 1 and 10 Hz.

3. If, at some point during the test, the specimen had stabilized to such an extent that strain control would approximate Neuber control, it was possible to change to strain control without stopping the test. The computer, on command, would complete one additional block under Neuber control, store the strain values of each reversal point, and then proceed through all subsequent blocks under strain control. This allowed the maximum ram displacement rate to be increased about 50 percent.

4. The computer directed a teletype that printed the reversal number, the stress, and the strain for every reversal point. High strain rates or small reversal amplitudes caused the teletype to fall behind in its printout. In this case, the computer stored the values in blocks of 50 consecutive reversals to be printed out later. However, the teletype did not print out the values of the reversals between blocks.

Sliding Line Control

The relative unavailability and expense of digital computer controlled testing systems made it desirable to develop a simple method of applying Neuber control. Effort in this direction has led to an approximation of Neuber control called sliding line control.

With this control technique, the Neuber hyperbola is approximated by a straight line on a stress-strain plot. The slope of the line (ratio of σ/ϵ) is the same for each reversal, but the intercept depends on the nominal

stress on the simulated notched member. A control circuit, which is presented in the Appendix, compares a signal proportional to the instantaneous value of the load history on the notched specimen being simulated with the appropriate linear function of stress and strain on the smooth specimen. The difference is integrated with respect to time and used as the strain command signal to the test system.

The advantages of sliding line control over digital computer control are:

1. It does not drift since stress and strain are always referred to the original zero axis.
2. The complexity of the load history is unlimited. For example, the system could simulate notch root behavior given a magnetic tape field record of the load history on a notched part.
3. Sliding line control is not subject to distortion from a series of small amplitude steps comprising a large cycle as explained by Tucker[8].
4. The frequency response is adequate to about 8 Hz.
5. The control circuit is external to the closed loop testing system.
6. It is inexpensive to equip an existing closed loop mechanical testing system with sliding line control.

The differences in strain between Neuber control and sliding line control are less than 0.0015 for a strain range of 0.04 and are zero for any elastic history. In a simulation where the load history consists of a high frequency low amplitude wave superimposed on a low frequency high amplitude wave (for example, a sine wave with noise), Neuber control develops errors and sliding line control does not.

Experimental Program

Materials and Specimens

Two types of specimens were used, notched plates and smooth cylindrical specimens. Details of the specimens are given in Fig. 3. The value of K_t for the notched specimens ranged from 1.75 to 2.0, depending on the material.

The four aircraft metals tested were 2024-T4 and 7075-T6 aluminum, heat treated aircraft quality SAE 4340 steel (with an ultimate tensile strength of 180 ksi) and Ti-8Al-1Mo-1V. Tensile and low cycle fatigue properties of these four metals may be found in Ref 9.

Test Equipment: All specimens were tested in an MTS closed loop, servo-controlled axial loading hydraulic test system. The stress and strain were continuously recorded. Strains on the smooth specimens were measured by a one-half-in. clip gage. Stress-strain hysteresis loops of the smooth specimen simulations were recorded on an X-Y plotter. Wood's metal grips were used to reduce alignment errors[10].

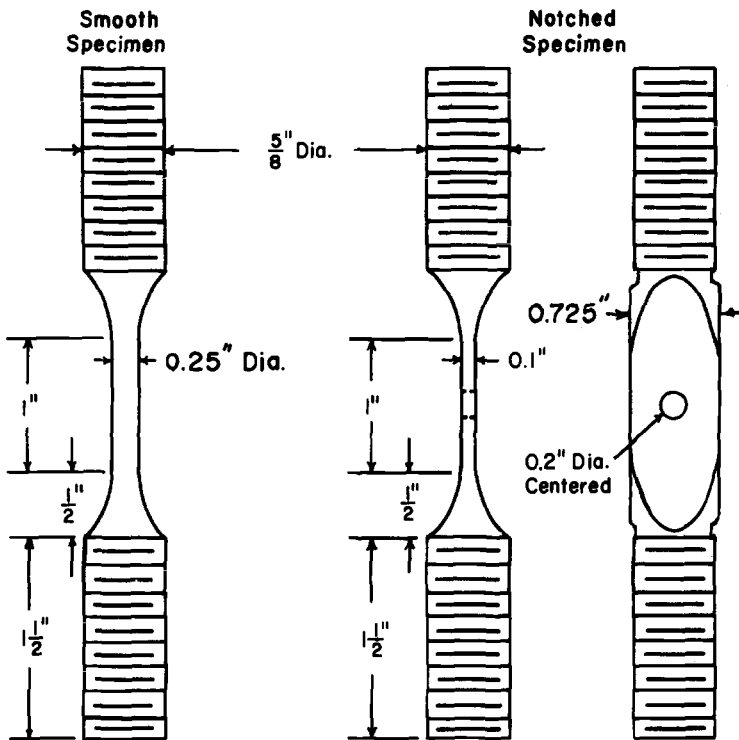


FIG. 3—Test specimens.

Test Program

Most of the experimental program consisted of the following two series of tests for each metal:

1. Notched members were subjected to an initial large cycle of loading followed by cyclic loading at a lower level until fatigue failure occurred. The magnitude of the first load was the same for all tests of the same metal but the sign was changed so that some were started in tension (load history *A* of Fig. 4) and other tests were started in compression (load history *B* of Fig. 4). These tests were designed to produce residual stresses at the notch root.

2. Smooth specimen simulations of the notch root stress-strain and fatigue behavior were performed for each of the notched specimen tests.

In addition, a few zero-maximum tension fatigue tests were performed and a series of notched specimens were subjected to the pseudo random load history shown in Fig. 5. For the pseudo random load history, notch root behavior was simulated on smooth specimens using sliding line control.

Failure is defined as the separation of the specimen into two pieces. These simulations are designed to predict only the cycles to crack initia-

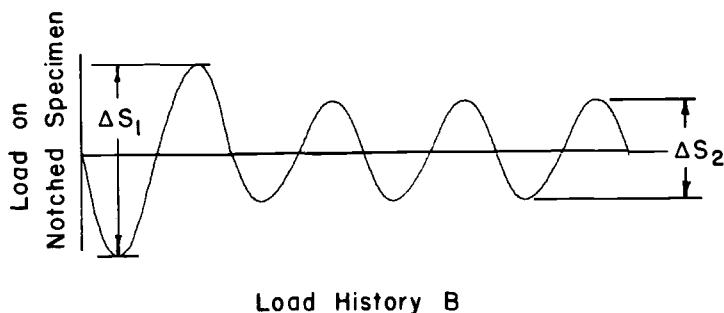
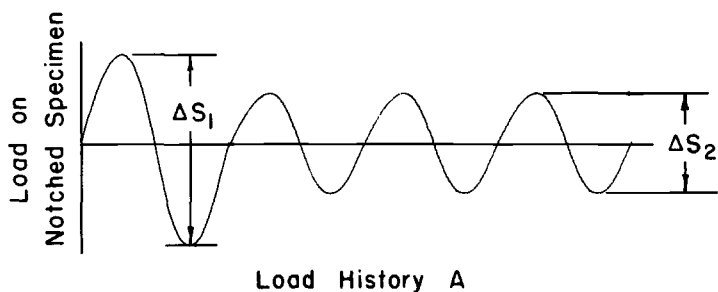


FIG. 4—Typical load—time sequence.

tion. In the tests conducted, the specimens were small and the metals were relatively high in strength, so that the proportion of the fatigue life devoted to propagation of an observable crack was a small percentage of the total life[11]. In applications where more ductile metals or larger specimens are used, or where the applied stresses are large, crack propa-

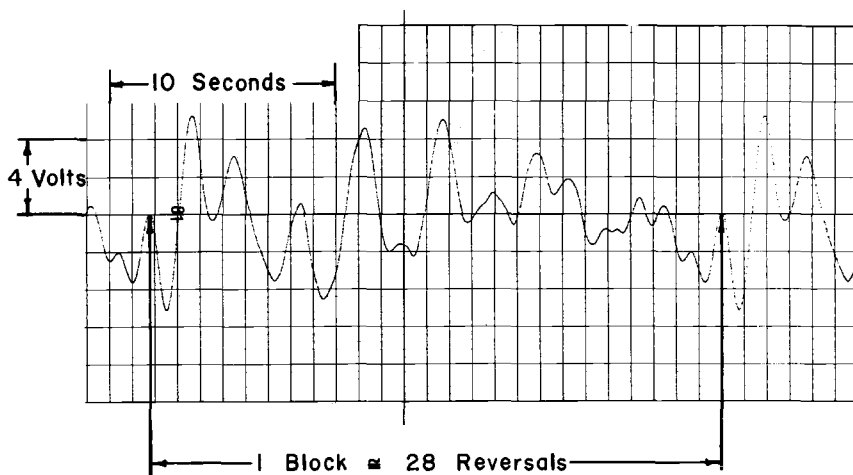


FIG. 5—Pseudo random load history.

gation life could sometimes be large in comparison with the crack initiation life and must be considered, especially in strain control situations.

Cyclic Stress-Strain Behavior at a Notch

The primary usefulness of a smooth specimen simulation is its ability to duplicate notch root stress-strain behavior. It is evident from the stress-strain plot of a smooth specimen simulation the way in which the local behavior affects the fatigue life of notched specimens. Since the smooth specimen undergoes approximately the same stress-strain history as the notch root metal, both cyclically harden or soften and develop or relax mean stresses in the same manner. The smooth specimen simulation allows the observer to estimate in a realistic way the stress range and mean stress at a notch root. These quantities, which are nearly impossible to obtain directly at the notch root, have been generally unavailable in the past and are responsible for widespread misunderstanding of a variety of sequence and overload effects.

Earlier attempts to analyze the notch root behavior, by those with an appreciation of the importance of the local stress and strain in relation to fatigue, have been limited by available information and techniques. Smith[12] has presented a method for estimating the residual stresses induced at a notch by an overload using the static stress-strain curve. Crews and Hardrath[13] and Crews[14] have measured strains at the root of large notches and forced the same strains onto smooth specimens to gain understanding of the stresses induced at the notch. Many industrial groups have developed their own methods for analysing the stress and strain at a notch and their effect on fatigue. Impellizari's paper[2] is an example. Martin et al[15] have recently presented a computer based fatigue analysis of notched members which requires cyclic stress-strain properties and the fatigue resistance of the metal as input.

In the approaches just reviewed, the analyst is forced to use static or an approximated stress-strain response of the metal. Most methods are generally unable to follow the cycle to cycle changes in the stress-strain properties and in the mean stress. In addition, the analyst is forced to adopt a convention of "cycle counting" and to assume some cumulative damage criteria incorporating the effect of mean stress and prestrain[16] to make his analysis relevant to fatigue failure.

Smooth specimen simulation circumvents most of these problems. The specimen acts as a "computer" which contains in its memory complete materials properties including its resistance to fatigue. Thus, the stress-strain response that is observed, and the fatigue life which is obtained, is not subject to assumptions concerning the material behavior. Rather, the assumption is made that the smooth specimen is representative of the notch root material, and that Neuber's rule, using the fatigue notch factor, K_f , is a reasonable way to relate nominal and local stresses and

strains. Admittedly, there are some features which are not perfectly simulated. For example, the state of stress may not be exactly uniaxial. Also, the volume of highly stressed metal which can have a significant effect on fatigue[17,18] is generally smaller in a notched member than in a smooth specimen.

Despite the lack of complete similitude, the use of smooth specimen simulations to recreate approximate local stress-strain histories in notched members can be enlightening as illustrated below.

Zero to Maximum Loading

Figure 6 is an actual X-Y recording of a smooth specimen computer controlled simulation test. It represents the local behavior of a notched 2024-T4 aluminum specimen repeatedly loaded from zero to maximum tension. Note that for continued cycling the mean stress relaxes and the material cyclically hardens. Thus, the tensile peak stress and strain remain about constant, while the locus of zero load stresses increases in compression with cycling, approximately describing the indicated portion

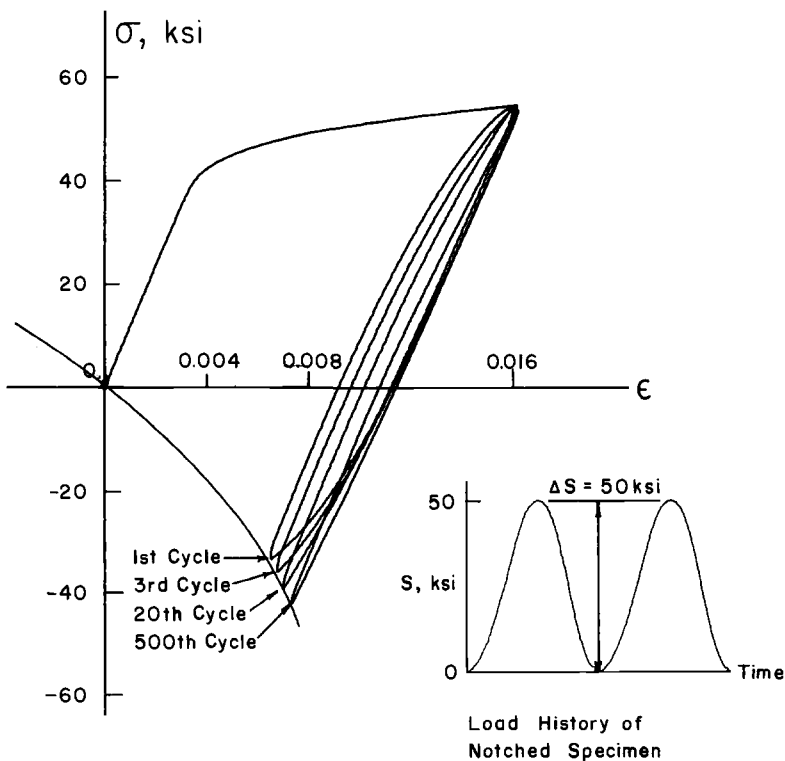


FIG. 6—Cyclic hardening and relaxation of mean stress under Neuber control (2024-T4 aluminum, $K_t = 2.0$).

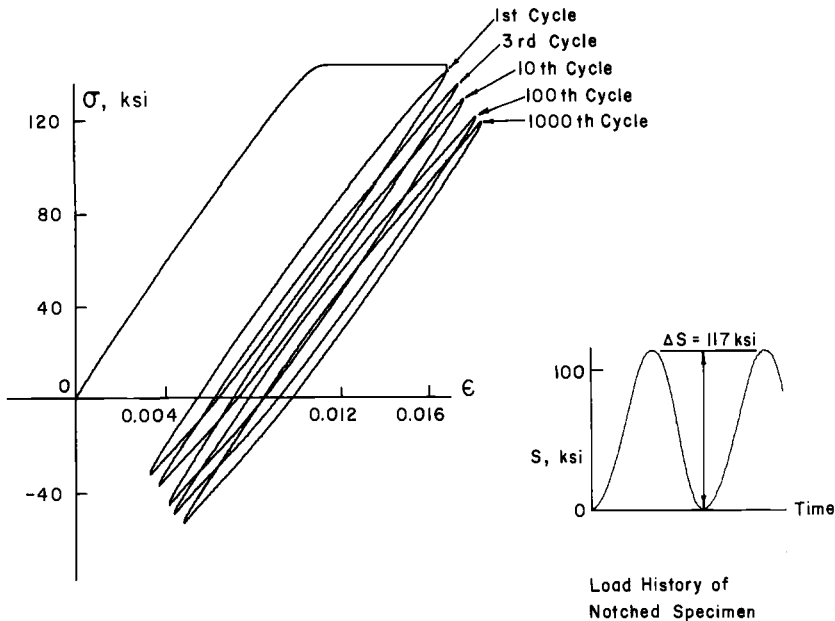


FIG. 7—Cyclic softening and relaxation of mean stress under Neuber control (Ti-8Al-1Mo-1V, $K_t = 1.75$).

of a hyperbola. This behavior is similar to the residual stress buildup at a notch root during zero to maximum tensile loading.

Figure 7 is a smooth specimen simulation of zero to maximum tension loading of a notched member of Ti-8Al-1Mo-1V. In this alloy the mean tensile stress also relaxes with repeated loading, but in a substantially different manner than for the 2024-T4 shown in Fig. 6. Only a small amount of cyclic softening is evident, but the maximum stress decreases with cycles while the maximum strain increases. The buildup of "residual stresses" at zero load is similar to that for the 2024-T4. Note that the locus of tips of the loops in Fig. 7 fall nearly onto a straight slanted line. Results such as these suggest that sliding line control is a reasonable approximation to Neuber control.

Sequence Effects

A smooth specimen simulation can be of great use in understanding the effect of loading sequence on the fatigue life of notched members. Dowling[19] has shown that most sequence and block size effects in smooth specimens can be explained in terms of the stress-strain response induced in the metal. A smooth specimen simulation of a notched member reflects all of the appropriate material responses.

For example, the load histories in Fig. 4 each contain one initial overload cycle, but the effects of the induced local mean stresses can affect

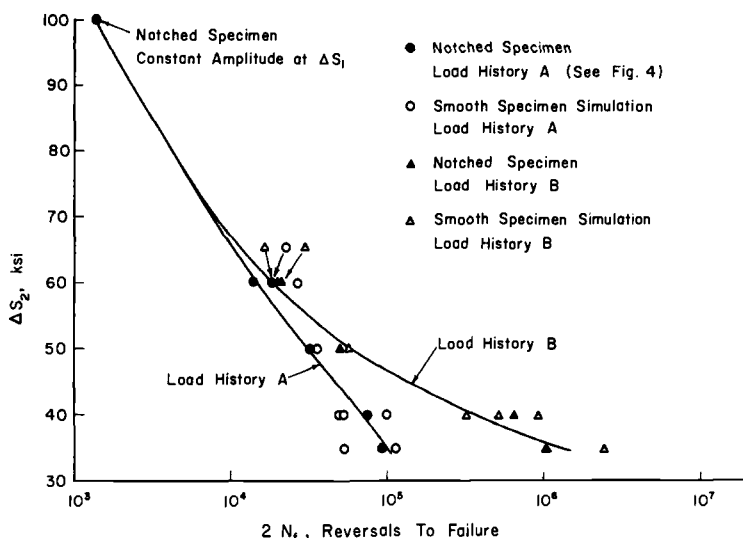


FIG. 8—Life as a function of S_2 for notched specimens and smooth specimen simulations (7075-T6 aluminum, $K_t = 1.95$).

the subsequent fatigue life by as much as a factor of one hundred (see Figs. 8 and 9). The reason for this large difference in fatigue lives is obvious from the X-Y plots of smooth specimen simulations shown in Figs. 10 and 11. As would be expected, these plots illustrate that load histories A and B result in mean stresses of opposite sign at the notch root.

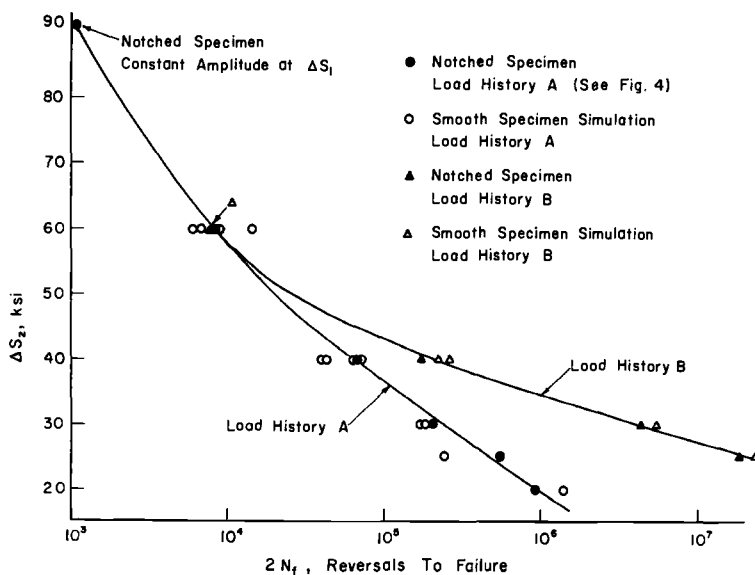


FIG. 9—Life as a function of S_2 for notched specimens and smooth specimen simulations (2024-T4 aluminum, $K_t = 2.0$).

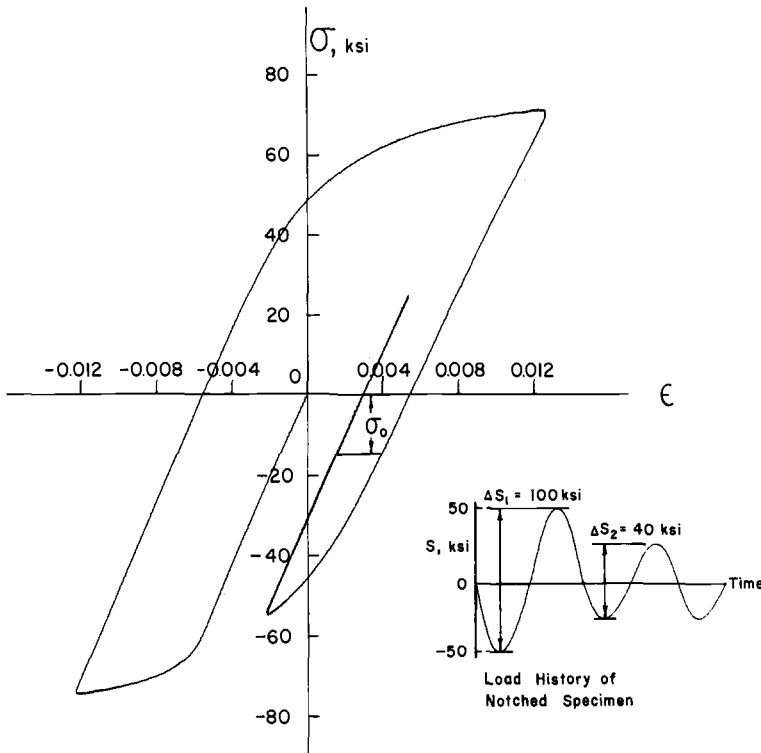


FIG. 10—Stress-strain response of a smooth specimen notch root simulation (7075-T6 aluminum, $K_t = 1.95$).

For the overload tests of Fig. 4, if the last large peak is tensile, the metal at the notch root is extended plastically. When the nominal load on the specimen is returned to zero, there is a local compressive residual stress at the notch root which will increase the fatigue life. If the last large peak is compressive, the local residual stresses are tensile and reduce the fatigue life. After the first cycle, the nominal stress level is reduced to a lower fixed amplitude for the remainder of the test. If this lower level is small enough, the metal at the notch root behaves essentially elastically. If the level of loading is high, the mean stresses relax toward zero due to cyclic plastic strain, and the mean stresses that are present after the initial preload make little difference in life.

Fatigue Results

Tables 1 through 5 give a comparison of the fatigue lives for a series of notched specimens and smooth specimen simulations. Tables 1 through 4 are for the load histories shown in Fig. 4. The notched specimens were tested under load control, and the smooth specimen simulations were performed under manual control, digital computer control, or

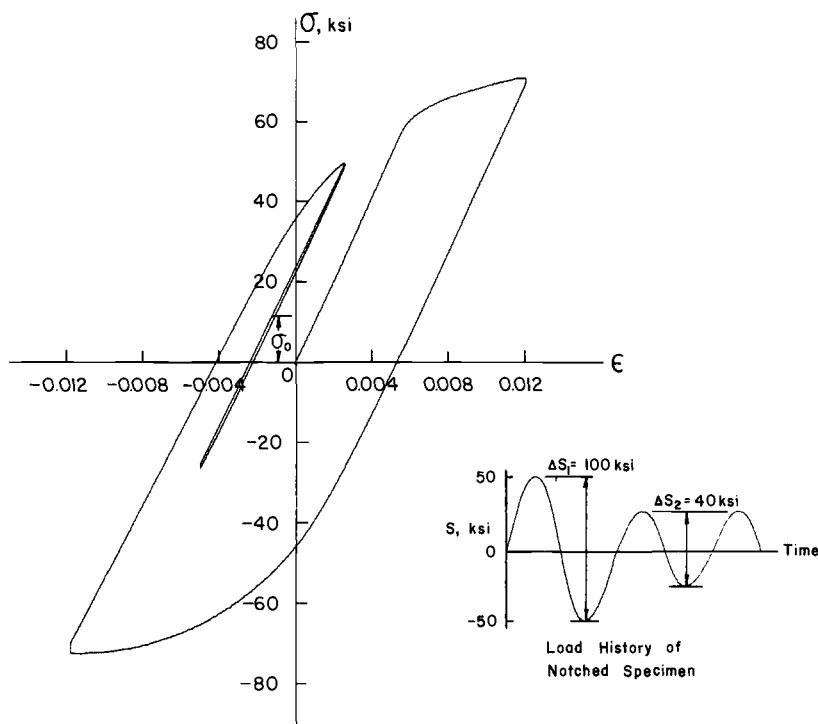


FIG. 11—Stress-strain response of a smooth specimen notch root simulation (7075-T6 aluminum, $K_t = 1.95$).

TABLE 1—Notched and smooth specimen life comparisons for 7075-T6 aluminum ($K_t = 1.95$)

Load History Type (Fig. 4)	ΔS_1	ΔS_2	Specimen Type	$2N_f$	Control
A	100	60	Notched	1.4×10^4	Load
A	100	60	Notched	1.2×10^4	Load
A	100	60	Smooth	2.7×10^4	Digital Computer
A	100	60	Smooth	1.4×10^4	Digital Computer
B	100	60	Notched	2.1×10^4	Load
B	100	60	Notched	1.9×10^4	Load
B	100	60	Smooth	2.1×10^4	Digital Computer
B	100	60	Smooth	1.8×10^4	Digital Computer
A	100	50	Notched	3.2×10^4	Load
A	100	50	Smooth	3.6×10^4	Digital Computer
B	100	50	Notched	5.1×10^4	Load
B	100	50	Smooth	5.8×10^4	Digital Computer
A	100	40	Notched	7.4×10^4	Load
A	100	40	Smooth	5.1×10^4	Digital Computer
A	100	40	Smooth	9.7×10^4	Sliding Line
A	100	40	Smooth	4.8×10^4	Sliding Line
B	100	40	Notched	6.3×10^5	Load
B	100	40	Smooth	9.1×10^5	Digital Computer
B	100	40	Smooth	3.1×10^5	Sliding Line
B	100	40	Smooth	5.2×10^5	Sliding Line
A	100	35	Notched	9.1×10^4	Load
A	100	35	Smooth	5.2×10^4	Digital Computer
A	100	35	Smooth	11.4×10^4	Sliding Line
B	100	35	Notched	1.0×10^6	Load
B	100	35	Smooth	2.4×10^6	Sliding Line

TABLE 2—*Notched and smooth specimen life comparisons for 4340 steel ($K_t = 1.85$)*

Load History Type (Fig. 4)	ΔS_1	ΔS_2	Specimen Type	$2N_t$	Control
A	209	82	Notched	3.0×10^5	Load
A	209	82	Smooth	3.8×10^5	Manual
B	209	82	Notched	2.4×10^5	Load
B	209	82	Smooth	4.5×10^5	Manual
A	209	73	Notched	3.2×10^5	Load
A	209	73	Smooth	4.2×10^5	Manual
A	209	73	Smooth	5.8×10^5	Digital Computer
A	209	73	Smooth	6.1×10^5	Sliding Line
A	209	73	Smooth	^a Did not fail	Digital Computer
B	209	73	Notched	^a Did not fail	Load
B	209	73	Smooth	^a Did not fail	Manual
B	209	73	Smooth	^a Did not fail	Manual
A	209	68	Notched	1.2×10^6	Load
A	209	68	Smooth	20.0×10^5	Manual
A	209	63	Notched	^a Did not fail	Load
A	209	63	Smooth	^a Did not fail	Manual

^a $2N_t > 10^7$ reversals.TABLE 3—*Notched and smooth specimen life comparisons for Ti-8Al-1Mo-1V ($K_t = 1.75$)*

Load History Type (Fig. 4)	ΔS_1	ΔS_2	Specimen Type	$2N_t$	Control
(0-Max.)	117	117	Notched	8.5×10^3	Load
(0-Max.)	117	117	Smooth	18.0×10^3	Digital Computer
A	216	153	Notched	4.6×10^3	Load
A	216	153	Smooth	4.8×10^3	Digital Computer
A	216	117	Notched	2.4×10^4	Load
A	216	117	Smooth	2.2×10^4	Digital Computer
B	216	117	Notched	2.5×10^4	Load
B	216	117	Smooth	1.8×10^4	Digital Computer

sliding line control, as is indicated by the last column. Table 5 gives a comparison of the notched and smooth specimen simulation lives for the more complicated load history given in Fig. 5. All of the tests in Table 5 were performed under sliding line control.

A review of the tables shows that the smooth specimen simulations predict the fatigue lives of the notched specimens within about a factor of two for the various metals and load histories investigated.

Figure 12 is a histogram showing the distribution of the smooth specimen simulation fatigue life divided by the notched specimen simulation fatigue life for the results presented. Tests where neither the notched specimen nor the smooth specimen simulation failed are not included.

Some difficulty was encountered in accurately predicting fatigue lives greater than 10^5 cycles in 4340 steel where a very small difference in either mean stress or strain amplitude makes a large difference in fatigue life. With these exceptions, the correlation between notched fatigue life and smooth specimen simulations is good.

TABLE 4—Notched and smooth specimen life comparisons for 2024-T4 aluminum ($K_t = 2.0$)

Load History Type (Fig. 4)	ΔS_1	ΔS_2	Specimen Type	$2N_f$	Control
A	90	60	Notched	8.0×10^3	Load
A	90	60	Smooth	12.0×10^3	Manual
A	90	60	Smooth	6.4×10^3	Digital Computer
A	90	60	Smooth	5.8×10^3	Digital Computer
A	90	60	Smooth	9.0×10^3	Sliding Line
B	90	60	Notched	7.9×10^3	Load
B	90	60	Smooth	8.6×10^3	Manual
A	90	40	Notched	6.8×10^4	Load
A	90	40	Smooth	4.2×10^4	Manual
A	90	40	Smooth	3.9×10	Digital Computer
A	90	40	Smooth	6.8×10^4	Digital Computer
A	90	40	Smooth	6.3×10	Sliding Line
B	90	40	Notched	1.7×10^5	Load
B	90	40	Smooth	2.1×10^5	Manual
B	90	40	Smooth	2.6×10^5	Digital Computer
A	90	30	Notched	2.0×10^5	Load
A	90	30	Smooth	1.9×10^5	Manual
A	90	30	Smooth	1.8×10^5	Digital Computer
B	90	30	Notched	4.2×10^5	Load
B	90	30	Smooth	5.6×10^5	Manual
A	90	25	Notched	5.5×10^5	Load
A	90	25	Smooth	2.3×10^5	Manual
B	90	25	Notched	1.7×10^7	Load
B	90	25	Smooth	^a Did not fail	Manual
A	90	20	Notched	9.4×10^5	Load
A	90	20	Smooth	14.0×10^5	Manual

^a $2N_f > 10^7$ reversals.TABLE 5—Notched and smooth specimen comparisons for 7075-T6 aluminum using the pseudo random load history of fig. 5 ($K_t = 1.95$)

Scale Factor for Fig. 5	Notched or Smooth	Blocks to Failure	Control
1 V = 10.0 ksi	Notched	261	Load
1 V = 10.0 ksi	Smooth	215	Sliding Line
1 V = 9.5 ksi	Notched	420	Load
1 V = 9.5 ksi	Smooth	457	Sliding Line
1 V = 9.0 ksi	Notched	597	Load
1 V = 9.0 ksi	Smooth	660	Sliding Line

Conclusions

If a smooth specimen is subjected to the same constraint as the metal at the notch root in a member subjected to fatigue loading, it will exhibit similar cyclic stress-strain behavior as the notch root metal (hardening, softening, and the introduction and relaxation of mean stresses). Also, because of similar stress-strain histories, the smooth specimen simulation of the notch root metal should initiate fatigue cracks at approximately the same life as the notched member.

Using smooth specimen simulations of notched members bypasses the need for a detailed knowledge of the cyclic stress-strain and fatigue properties of the metal and it makes it unnecessary to adopt a cycle counting

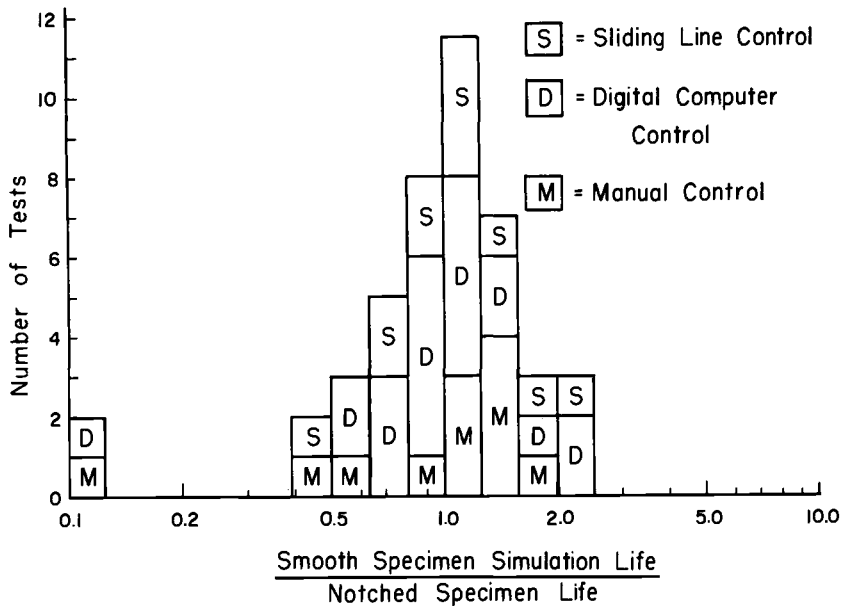


FIG. 12—Distribution of failure predictions.

method or to assume a fatigue damage criterion. All that is needed is a way to relate nominal stresses on the notched member to local stress and strain at the notch root. Several techniques for controlling the smooth specimen simulation according to Neuber's rule have been developed. They have been shown to provide an acceptable estimate of the life of notched members.

Acknowledgments

This investigation was conducted primarily in the H. F. Moore Fracture Research Laboratory of the Department of Theoretical and Applied Mechanics, University of Illinois in cooperation with the Naval Air Development Center, Warminster, Pa. The digital computer controlled tests were performed at MTS Systems, Corp. of Minneapolis, Minn. and at the Department of Civil Engineering, University of Waterloo, Waterloo, Ontario, Can.

We are greatly indebted to R. E. Skoe of MTS Systems, Corp. who wrote the digital computer control program, and to A. Conley of the University of Waterloo and N. E. Dowling who provided advice and technical assistance.

APPENDIX

Sliding Line Control by S. J. Stadnick

Sliding line control is the result of an attempt to simplify Neuber control by replacing the stress-strain hyperbola with a straight, slanted line (Fig.

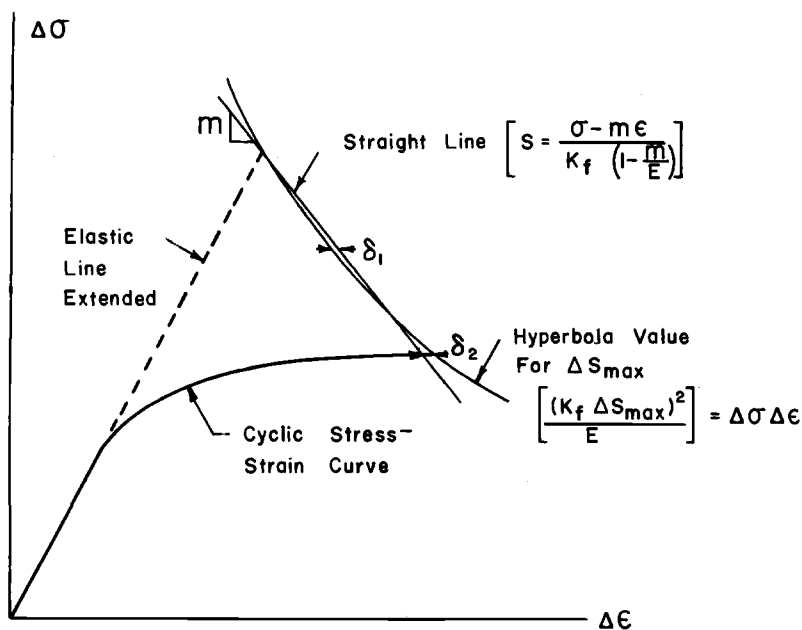


FIG. 13—Slanted line approximation to Neuber hyperbola.

13). The straight line is represented in terms of local stress and strain as:

$$S = \frac{\sigma - m\epsilon}{K_f \left(1 - \frac{m}{E}\right)}$$

where

S = value of the nominal stress on the notched specimen being simulated as a function of time,

σ = stress on the smooth specimen,

ϵ = strain on the smooth specimen,

K_f = fatigue notch factor, and

m = slope of the sliding line.

It is helpful in understanding this type of control to consider the simplest of special cases—a single reversal wherein the strains are essentially elastic. In any reversal where the material is elastic, $\epsilon = \sigma/E$, and substituting into the equation of the sliding line,

$$S = \frac{\sigma - \frac{m\sigma}{E}}{K_f \left(1 - \frac{m}{E}\right)}$$

or

$$S = \frac{\sigma}{K_f}$$

Note that the magnitude of change in stress and strain in response to an entirely elastic reversal does not depend in any way on the previous load history,

previous plastic strain of the notch root simulation, or the slope m of the sliding line. The magnitude of change in stress at the notch root for elastic stress-strain response is always $\Delta\sigma = K_f \Delta S$ without the rezeroing necessary in Neuber control.

The slope of the line is important for maintaining a close approximation to Neuber control for reversals where there is significant plastic strain in the notch root simulation. Two conditions determine the slope of the slanted line. First, the solution of Neuber's rule and sliding line control must be identical in the elastic case (see Fig. 13). With the intersection of the elastic line and the Neuber hyperbola as a given point, the slope m of the straight line is determined to provide a "best fit" of the straight line and the Neuber hyperbola for the maximum range of local strain expected in the course of the test. The maximum expected strain is found by estimating the maximum and minimum nominal stress limits during the entire load history on the notched specimen, and defining the difference between them as ΔS_{max} . Then, calculate the value of a Neuber hyperbola using the value of ΔS_{max} . A cyclic stress-strain curve superimposed on the hyperbola gives a good indication of the maximum strain. The slope is not very sensitive to small differences in expected strain, so exact values of maximum strain are not necessary.

Sliding line control does not drift because the position of the line is a linear function of the stress on the notched specimen, and contains no quantities dependent on the points of reversal or the previous stress-strain history.

A careful reversal by reversal analysis of the relationship between sliding line control and Neuber control for different notched specimen load histories has shown that the maximum differences between the two control conditions is approximately $\delta_1 + \delta_2$ of Fig. 13 for most cases. The one exception is for a load history consisting of a high frequency low amplitude wave superimposed on a high amplitude, low frequency wave. For this particular history, sliding line control is representative of the notch root metal and Neuber control, as defined in this paper, predicts far more plastic strain at the notch root than actually occurs.

Analog Control Circuit

Figure 14 is the circuit used to enforce sliding line control on a smooth specimen. A flow chart is presented in Fig. 15. The circuit is similar to most analog computer controlled circuits except that the difference between the command and feedback is not used directly to control the movement of the ram, as in Slot et al[20], but is integrated and the integral used to control the ram in strain control. This form of control is somewhat similar to that proposed by Richards and Wetzels[21], but is more versatile and can be used for any input load history, not just constant amplitude.

This modification of a standard analog computer control circuit is desirable for several reasons:

1. The control circuit is entirely external to the closed loop testing system. The output of the integrator is connected to the input of the strain control, just as if it were a function generator. No internal modifications or connections are needed in the closed loop system.
2. The input to the closed loop system may be modified in a variety of ways. For example, a pair of Zener diodes from the output of the integrator to ground would limit the strain input to the breakdown voltage of the diodes. This would prevent the test system from ever losing control because of a too large or accidentally incorrect input voltage. A safety feature of this

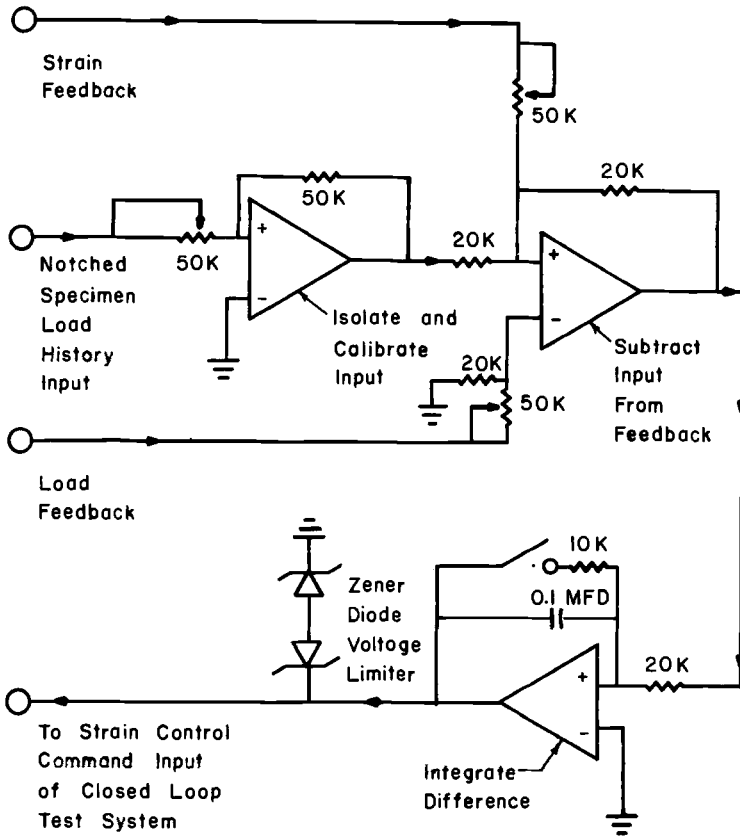


FIG. 14—Schematic of sliding line control circuit.

type considerably reduces the anxiety of performing tests with a new type of control circuit.

3. This same circuit may be used to control true stress, true strain, plastic strain, or any other quantity which is determined by an analog feedback circuit.

An analysis of the flow chart (Fig. 15) provides insight into the closed loop nature of the control circuit. Under sliding line control, an increase in the value of S causes the difference between the voltage value of the sliding line and the voltage value S to differ slightly from zero. This difference is then integrated with respect to time. Since the integrated value of the difference is the command input for strain control to the test system, a change in the integral value causes a change in strain in the specimen. This causes the voltage value of the sliding line to change. The net result, providing that the phase of the external closed loop is correct, is that the voltage value of the sliding line as determined by the feedback will follow closely the voltage value of the load history, regardless of its complexity.

If the phase is not correct, the value of the integral will quickly pin itself at the limit allowed by the Zener diodes. Phase can then be corrected by inverting the output of the integrator.

An important factor in the response of the circuit is the time constant of

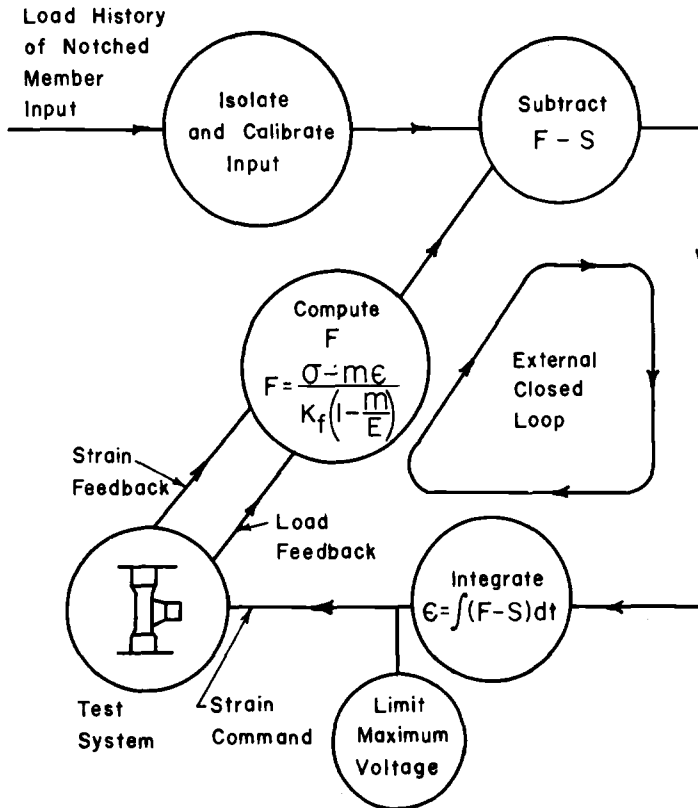


FIG. 15—Flow chart for sliding line control.

the integrator. If it is too long, the rate of change of the integral will be slowed so that it cannot follow the input signal closely. If the time constant is very short, the circuit has a tendency to oscillate. The best results so far have been with a time constant of about $\frac{1}{30}$ that of the command signal frequency. The lag with this constant is less than 0.1 percent of the signal value. The circuit presented is by no means optimized in any way. It performs sliding line and other forms of feedback control adequately, but certainly can be improved upon.

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Apollo Quality Through Predictive Testing

REFERENCE: White, G. C., Jr., "Apollo Quality Through Predictive Testing," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 253-267.

ABSTRACT: Successful operation of a machine by a man is contingent upon meeting the following condition: the capability of the machine must be compatible with the capability of the man in the environment which they will see together. How do we achieve this compatibility? Through testing of hardware and training of people in the best simulation of operational environment that is economically feasible. This paper discusses the testing performed during the Apollo program and its relationship to the Engineering, Qualification, Manufacturing, Maintenance and Training aspects of the program. Illustrative examples of Apollo experience are used to highlight the discussion.

KEY WORDS: predictions, machine design, environment simulation, environmental engineering, structural loading, inspection, preventative maintenance, manufacturing, contingency, coordination, engineering drawings, process control, personnel development

Successful operation of a machine by a man is contingent upon meeting the following condition: the capability of the machine must be compatible with the capability of the man in the environment which they will see together. This principle holds through broad spectra of mechanical complexity, operator skill, and environmental factors: from a woman with her washing machine in the ambient environment of her kitchen, to a man driving a car on a congested high-speed highway, to an astronaut in a moon-bound Apollo space vehicle in the demanding conditions of space.

How do we achieve this compatibility? Through testing of hardware and training of people in the best simulation of operational environment that is economically feasible. This principle can be thought of as a closed loop system as depicted in Fig. 1. I will discuss the general relationships of the functions shown in this figure and then consider each one in more depth, citing some examples to illustrate their importance and bringing out pertinent test requirements. Finally, I will discuss some related fac-

¹ Deputy director, Office of Manned Space Flight Reliability, Quality and Safety, National Aeronautics and Space Administration, Washington, D. C. 20546.

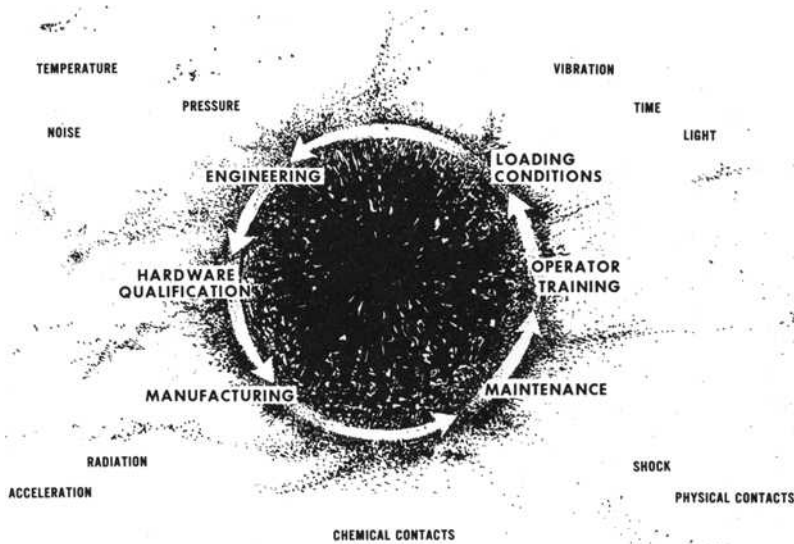


FIG. 1—*Man-Machine-Environment relationships.*

tors which do not appear on this chart but which are nevertheless of paramount importance in the total scene.

The usual starting point, after deciding on a general design concept, is the establishment of Design Loading Conditions. In a broad sense this can be thought of as structural loading, electrical loading, pressure loading, and all other forms of stress which the machine will see, including its test and operational environments.

The total Engineering task is then accomplished including analysis, preparation of drawings and specifications, parts and materials testing, and development testing of components, subsystems and the completely assembled article. This work must, of course, be based on the established loading conditions and must produce a design which can be manufactured, inspected, and operated with confidence that objectives will be achieved within the established constraints.

Hardware Qualification consists of testing at an appropriate assembly level of hardware identical in configuration and manufacturing process to the operational hardware. Test conditions should represent worst probable combinations of loading conditions and environments, and for some critical hardware should include off limits testing to determine margins of safety. Basic objective of qualification testing is proof that the hardware meets specification requirements.

During the Manufacturing phase one of the primary considerations must be the maintenance of processes, tools, and operator skills consistent with those used in manufacturing the qualification test article. In-process inspection and testing, as well as acceptance testing of the

completed article are performed to give assurance that this consistency continues.

Maintenance in the field usually includes some manufacturing as well as inspection and modifications. It is important to keep in mind necessary differences in field installation from factory installation brought about by different tooling, processes, and working conditions. Where these differences affect critical hardware, qualification testing should be conducted to prove the acceptability of the differences.

To close this loop and ensure that the manufactured hardware is not asked to exceed its design loading conditions, operators must be trained. This includes acquiring knowledge of the design conditions, developing the ability to assess current conditions through instrumentation or personal observation, learning how to control the machine to stay within design conditions, and acquiring the self discipline to apply this knowledge continuously. Contingency conditions must also be hypothesized, and procedures developed and practiced so that the operator can recognize and react to such conditions if they arise.

Thus far, this discussion has been sufficiently general to apply to all levels of complexity—the washing machine, the automobile, the space vehicle. Now, the areas discussed above will be covered in more depth, using experience gained in the Apollo program as illustrations. I do not intend to go into all details of the six points in the loop, but rather will bring out examples of factors which are related to potential problems, or which may be more important than is at first apparent.

Loading Conditions

Primary Loads

Determination of the loading conditions on a vehicle is actually a part of the Engineering process, but in order to bring out its relation to Training and Manufacturing I have chosen to discuss it as a separate item.

The primary loading conditions are established on the basis of past experience, analysis, and tests. Past experience is to some degree reflected in regulations, codes, specifications, and standards for aircraft, ships, and rail and highway vehicles. This experience includes years of trial and error in manufacturing and operations and tends to tell the designer what he can anticipate in the way of manufacturing technicians' skills and pilots' capabilities. How many G's will a crane operator apply to a wing hoist fitting or how many G's will a pilot cause to be applied to a transport aircraft structure? The knowledge and self discipline of the operators are in each case the major determining factors, and years of experience have resulted in established methods for determining such loads.

Where directly related past experience is not available, the designer

resorts to analysis of whatever information is available. He hypothesizes all possible conditions, determines to the best of his ability the effect of human inputs, and then conducts tests of the vehicle to confirm or modify his hypothesis.

An example of this procedure is found in the development of the Apollo Lunar Module landing gear. Best estimates of lunar soil consistency were based on photographs and radar data and later confirmed by Surveyor landings. This determined the size of the foot pads.

A $\frac{1}{6}$ scale model was dropped on surfaces at various slopes and with simulated bumps under several possible combinations of forward, side-ward, and vertical velocities. These tests provided a basis for loading conditions used in structural design of the landing gear. Fig. 2 depicts a drop test of this $\frac{1}{6}$ scale model.

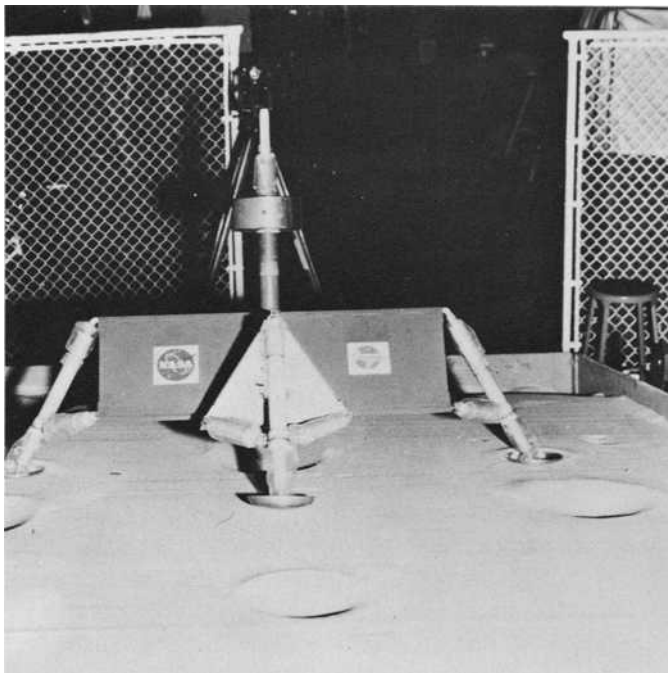


FIG. 2.—Drop test of $1/6$ scale Lunar Module model.

Contingency Conditions

Under contingency conditions, that is, after a failure of a critical component, loading on remaining systems or components is sometimes increased and this must be taken into account in the design.

Illustration of this consideration of contingency loading conditions is provided by the design of parachutes. Here drop tests of the complete earth landing system with a flight weight Apollo Command Module have

proven that if one of the two drogue chutes fails and one of the three main chutes fails, the remaining chutes will provide sufficient deceleration and are strong enough to result in a safe landing.

Manufacturing Loads

A category of loading which is easy to overlook is the loading encountered in the various phases of manufacturing, including assembly. While major loads such as hoisting of a launch vehicle stage are usually given adequate attention, lesser loading such as is imposed on components by using them as handholds, or loads imposed by test equipment which are different from the flight design loads might be ignored or given too superficial an analysis.

A number of cases of leakage occurred in soldered tubing joints early in the program where tubing was joined by soldering the ends into a sleeve. Part of the problem here was that although the joints were sealed and could withstand anticipated vibratory loading, they were not strong enough in bending and torsion to carry the relatively light loads imposed during installation and normal bumps and pressures exerted by technicians working around them. This problem was solved by installing a two-piece metal sleeve around the soldered joint potted with epoxy to strengthen the joint in bending.

Similar problems have occurred with welded joints in plumbing. Electrical wiring harnesses are very susceptible to damage by installation loads and inadvertent abrasion by workers. Adequate room must be provided for installation, temporary or permanent protective guards must be provided over tubing and wiring, and technicians must be impressed with the susceptibility to damage of these types of components.

Engineering

Non-Engineering Inputs

Failure of the design engineers to obtain the collaboration of manufacturing, maintenance, quality control, test, and operations personnel during the design phase may result in overlooking some important design features related to these functions. Furthermore, developing a team effort involving these people early in the game results in determination to make the hardware work rather than merely blaming their problems on the "stupid engineers."

In the manufacturing area, a problem arose on the Apollo program from the existence of "look-alike" parts; that is, parts whose external appearance was identical but which had different wall thicknesses and, of course, different part numbers. Due to the fact that identification was lost during a vat processing operation, the parts were installed in the wrong location.

This incident was discovered through the serendipity which sometimes occurs in test programs, knowledge being gained which was not anticipated. In this case, a strut failed in test due to an error in internal machining. During the ensuing inspection of all struts of several different designs the misidentification of the "look-alikes" was found. The entire vehicle was then reviewed for look-alikes to insure against other wrong installations. Moral: design a difference in external appearance in potential look-alikes, maintain part identification throughout processing, draw all possible knowledge from every test, and act on the basis of that knowledge.

Coordination with flight crews is important from two viewpoints. Their experience and preferences can result in a more easily operated vehicle, while their participation in the design process adds to their knowledge of the hardware and assists them in developing sound operational procedures.

Communication between design engineers and test and quality control people in design phases is necessary to ensure that the produced hardware can be inspected and tested in the most effective manner. Designing of test points into electronic systems and bleed ports into hydraulic systems can preclude later more expensive changes or makeshift work-arounds. Provision for checkout of both legs of redundant systems is mandatory to be sure of having true redundancy.

Need for this coordination with test and quality control is greatest where materials or design concepts are utilized that have not fully matured in production and operation. In Apollo, we used aluminum alloys and high strength steels in fabricating pressure vessels where it was later found that operating conditions made them susceptible to the "fracture mechanics" type of brittle failure. We developed improved test and inspection procedures in real time to gain assurance that they were safe to operate. Additional tests included cryogenic pressure tests of the Saturn S-II stages until we were confident that inspection procedures could in fact detect a critical size flaw.

Drawings and Specifications

The mundane process of preparing and checking drawings and specifications creates the foundation on which the program is built. Manufacturing planning, tool design, test fixtures, test procedures, operations procedures, and crew training are all based ultimately on the information recorded on drawings and in specifications. If this information is inaccurate or incomplete, the way is opened for a future problem when the necessary contributing circumstances coincide.

One of the factors which may have contributed to the oxygen tank failure on Apollo 13 was the absence of a voltage requirement in a procurement specification for a switch. Although the switch was "off-the-

shelf," it was used under conditions somewhat different from the vendor's specification conditions. Under the normal circumstances prevailing at the time of the original design no failure occurred, but with the passage of time the voltage used during pre-launch filling and draining procedure was increased. This, coupled with a nonstandard pre-launch procedure necessitated by a loose plumbing joint internal to the tank, resulted in overheating of internal wiring and a subsequent short in flight. If the voltage had been called out in the specification, the inconsistency between the design and qualification test voltage and the pre-launch test voltage would have surfaced in document review.

Such a seemingly minor thing as poor quality reproduction of drawings can result in a barely legible number which may be misread by a technician making or installing a part or by an engineer designing a mating part. Control of changes to drawings and specifications and distribution of revisions is another routine but essential element of an effective engineering operation.

Development Testing

The term "development testing" is used here to include all the testing that is done as part of the engineering process up to the time when a suitable hardware design is complete and ready for qualification. It includes tests of parts and materials to determine their suitability, test of breadboard components or subsystems to prove a concept, and tests of multiple subsystems to investigate interactions.

In some cases the cost of test hardware and the cost of test operations are so great that development testing and qualification testing are merged. Data obtained are utilized as design information and yet the hardware is close enough to the production configuration that analysis of the effect of minor differences produces high confidence that the production hardware will meet specification requirements. Examples of this kind of test in the Apollo program are the thermal vacuum tests of the spacecraft in the altitude chamber at the Manned Spacecraft Center in Houston (Fig. 3) and the full scale vibration tests of the Saturn V vehicle stack at the Marshall Space Flight Center at Huntsville (Fig. 4).

Feedback

There is a strong tendency in some engineering departments to feel that their job is done when the drawings are released. In reality, the experience gained in manufacturing, testing, training, and early operational usage must form the basis for incorporating in the hardware those changes found necessary to preclude occurrence of problems in operation. Utilization of the feedback from these experiences is essential to successful hardware operation.

The Apollo program included a strong, formal program to report,

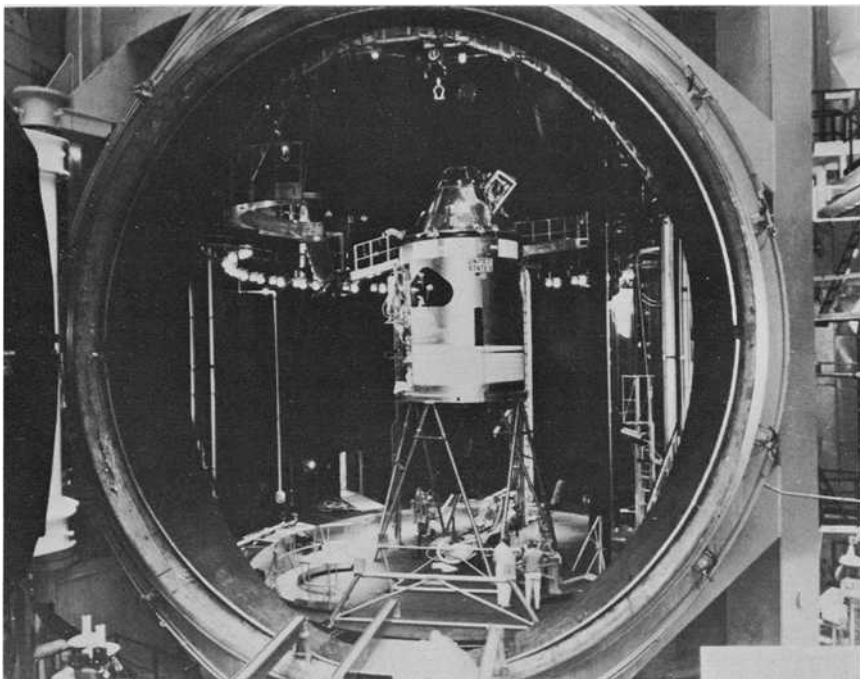


FIG. 3—*Command and service modules in thermal vacuum test chamber.*

analyze and correct problems. Critical nonconformances and flight anomalies were resolved prior to the next flight.

Qualification

Qualification testing is conducted to prove that the hardware is capable of meeting specification requirements. Ideally, it is performed on a piece of hardware that is identical to production hardware, not only in its final configuration but in the processes, tooling, and techniques used in producing it.

Actually, since qualification is done early in a program's history it is likely that engineering changes or processing changes during or subsequent to qualification will invalidate to some degree the qualification test results. Analysis is necessary to determine how much qualification testing should be rerun. In some cases, analysis is not entirely feasible and the judgment of a person, expert in that hardware, is used. For Apollo spacecraft, the examination of qualification test data, analysis of differences and expert judgment are combined in a formal process.

Assembly Levels

It is fairly well agreed that qualification testing of all hardware is necessary to produce an operationally satisfactory system. However, there

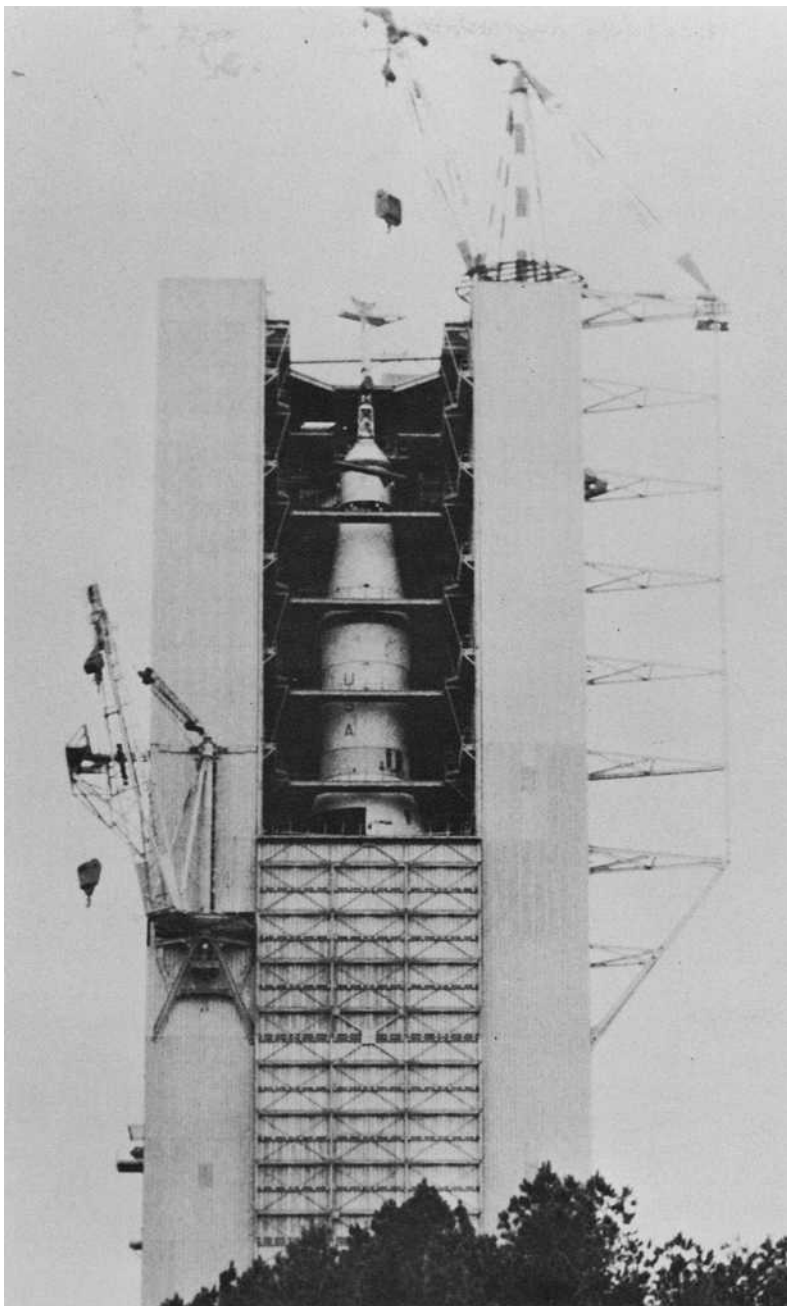


FIG. 4—*Saturn V dynamic test vehicle in dynamic test facility.*

is considerable difference of opinion as to whether this means testing of every piece part individually or testing of the completely assembled product or some combination of testing at various assembly levels.

In general, in the Apollo program electrical, electronic and electro-mechanical parts are qualified at the piece part level (transistors, connectors, switches). Assembled components such as motors, valves, and gyros are qualified at the component level. Selfcontained units of sub-systems such as the Fuel Cells, and the Environmental Control System power unit are qualified at this selfcontained unit level whereas the complete Electrical Power System and the complete Environmental Control System are qualified as a part of the Command and Service Module thermal vacuum tests.

The assembly level of qualification testing is usually decided on a judgmental basis in consideration of such questions as: What testing plan will result in the lowest total cost? What plan will produce the best simulation of operational conditions? What will be the effects of a failure under the various plans in terms of hardware damage, cost, and schedule slippage?

Fluid Qualification

One significant area of learning in the Apollo program was the need to qualify by test and control the fluids used in processing or operating the hardware. Nitrogen tetroxide is used as an oxidizer in the main service module engine and it is carried in titanium alloy tanks. After a failure of these tanks had been experienced during a pressure test using methanol as a test fluid, it was found that anhydrous methanol induced stress corrosion in the titanium alloy tank, whereas presence of a small amount of moisture inhibited this action. Similarly, it was found that pure nitrogen tetroxide induced stress corrosion in these tanks, while the presence of a very small amount of nitric oxide would inhibit the stress corrosion.

Manufacturing

Once the engineering is complete and the hardware has passed qualification testing, the reliability of the product is totally dependent on the integrity of the manufacturing operation. This means the ability to produce exactly what was designed and qualified.

Manufacturing Planning

Compatibility of the design with available manufacturing processes should have been established early in the engineering phase of the program by close collaboration of engineers and manufacturing people. Planning of manufacturing processes and selection or design of tooling and equipment must guarantee consistency in all details of the processing. Planning tickets or "travelers" are just as important as drawings and

must carry sufficient detail to permit the technicians to make parts and test them correctly and consistently throughout the program.

Process Control

Testing of processes is necessary to ensure initial adequacy and in some cases to ensure continuing adequacy. While this is done traditionally in such things as vat processing, it is sometimes not properly done with respect to mechanical operations, particularly in details that are thought of as "workmanship." The following examples illustrate this point.

It was found that in welding a small fluid line excessive drop-through had occurred, partially restricting the line. The technician used a small burring tool to grind out the drop-through material, and it was subsequently found in checking a small leak at the weld that this operation had dangerously reduced the wall thickness of the tubing in a localized area. In this case the welding and clean-out operations had not been standardized and approved to assure a satisfactory product.

A small switch had to be modified by grinding off certain portions of the plastic case to provide mounting clearance. During this operation a paraffin-base material was used to seal the gap around the plunger so that debris from the grinding operations could not get inside the case. This material was later removed. During checkout after installation it was found that switch action was sluggish and upon investigation, a trace of the paraffin-base material was found on the plunger. This material and process had not been qualified by test to prove their acceptability.

Maintenance

There are two aspects of Maintenance operations that I would like to emphasize in the context of this discussion: modifications which are sometimes accomplished as part of the maintenance procedure, and the need for feedback of information on problems encountered.

Modifications

Modification of hardware always carries the implication of change to something which has been qualified and possibly acceptance tested and checked out. It is assumed that any authorized modification has been previously qualified, but at the time of installation of the modification, consideration must be given to the need for repeating any acceptance or check-out testing which had been accomplished prior to this modification. This retesting may be necessary to check out the modified hardware itself, or its interaction with other parts of the system, or to check hardware which was removed to permit access and then reinstalled. Retesting may also be required after replacing age limited or cycle limited parts in routine maintenance.

In one of the launch vehicle stages used in the Apollo program a modification was made incorporating a new component in the system which provided for engine shutdown in case of a malfunction causing combustion chamber pressure to fall below a minimum limit. Installation of this component involved reconnecting of cables leading to the engines. Two of these cables were interchanged at their connectors and it so happened that one of these two engines did malfunction during the boost phase in an unmanned test flight. Due to the cable interchange, the shutdown signal went to a good engine which did in fact shut down. As this engine lost chamber pressure, it in turn sent a shutdown signal to the malfunctioning engine which also shut down. Although this event occurred late in the boost phase, it contributed to inability of the vehicle to achieve its development test objectives. In this case, the test run after the modification proved only that the system would operate satisfactorily in a mode which called for all engines to shut down and therefore it did not detect the error in individual engine hookup.

Feedback

Feedback of problem information to Engineering from a Maintenance activity is useful in minimizing failures and improving the reliability and serviceability of the product. Corrective action resulting from such feedback might take the form of a design change and hardware modification, a change in operating procedure, or initiation of a new or revised inspection or checkout procedure. Each case is reviewed by management, considering that requalification testing may be required, that changes to existing hardware may introduce new problems, and that the cost in time and money may not be justified by the potential improvement in serviceability.

Operator Training

Training of operators is the all important closure to this man-machine compatibility loop. This phase of the Apollo program probably represents the most thorough and comprehensive training of man for a specific task that has ever been accomplished.

Background Training

Training of the astronauts for an Apollo mission was built on an impressive foundation since each astronaut had military or test pilot experience as well as a science or engineering degree prior to his selection. Several years of living with and studying the Apollo spacecraft and launch vehicle and all their complex subsystems gives them a wealth of knowledge as to what happens when they flip a switch or what may cause an instrument reading to deviate from normal.

Simulators

Experience in various types of simulators gives the astronaut not only the knowledge but also the feel of various aspects of the lunar mission. A spacecraft simulator includes closed circuit television screens at the windows to reproduce the scenes that they would see in an actual mission. The lunar landing training vehicle simulates the dynamic responses of the Lunar Module as it simulates in actual flight an approach to a lunar landing.

Network simulations with the crew in the spacecraft linked into the worldwide communications network run through the routine of various crucial mission phases. Throughout these months and years of training they maintain flight proficiency by flying jet aircraft and helicopters.

Mission Problems

With all this emphasis on training, however, there are occasional minor incidents that remind us that the crews and their trainers are, after all, human; witness the damage to the television camera on the lunar surface when it was inadvertantly turned toward the sun. At the same time, recovery from the near catastrophe of Apollo 13 gave evidence of the value of the thorough background training of both the flight and ground crews. In this case the simulators were used to test emergency procedures devised on the ground prior to transmitting them to the crew.

Technician Training

Training of technicians conducting ground tests is another important activity, for although ground tests are not usually in the public eye as are flight missions, they nevertheless involve potential hazard to life and to millions of dollars worth of hardware. We have experienced partial collapse of large thin walled tanks (both flight vehicle and ground support) by drainage of fluid without adequate venting. In another instance a test launch vehicle stage was destroyed by over pressurization when a test crew was not informed by the previous shift that a pressure gage had been disconnected for trouble shooting.

Related Factors

Economics

Economic considerations come into play primarily in two ways with relation to a given test program; (a) during initial planning of the program the cost of facilities, test articles, and test performance are scrutinized closely for justification in terms of anticipated results, and (b) changes found necessary as a result of test performance are questioned as to their benefit in terms of reduction of risk or saving of future costs.

It is incumbent upon the test engineer to recognize this need on the part of management and furnish information in management's terminology which will assist in making the best decisions from a long range view.

Another aspect of the economics of test programs warrants mention. In recent years the physical size and monetary cost of test articles has rapidly increased, and at the same time public visibility of test results has also increased. The spectacular newsworthiness of failure of a large and expensive piece of hardware can draw charges of wastefulness from Congress and the press, although in many cases more can be learned in a test program from a failure than from nonfailure. This eventuality must be given consideration in test planning and we must make clear the value of any test designed to produce failure.

Motivation

Much attention has been given in recent years to the importance of motivating people to take pride in doing their work correctly while recognizing the need to meet realistic schedules. Paradoxically, the institution of rigid formalized procedures and double checking of every operation seems in itself to de-motivate people and de-emphasize the importance of pride in workmanship. However, where lives and large investments are at stake, there is no substitute for rigidly disciplined procedures.

In order to resolve this paradox the attitudes of management, both expressed and implied by their actions, must make clear the importance of individual workmanship and the usefulness of the disciplines in assisting each worker to turn out a good product.

Quality and Reliability Functions

There is frequently a reluctance on the part of Test Engineers to use Quality and Reliability people to help assure good test operations, particularly in "development" testing. This is understandable since it stems primarily from the engineer's pride in his ability to do the job properly. However, for the more expensive development tests and certainly for qualification, acceptance, and check out, the Quality and Reliability people can be of major help in assuring valid results by:

1. Reviewing test plans and procedures.
2. Inspecting the test article and test equipment prior to test.
3. Checking each step of the procedure as the test is performed.
4. Recording all failures and deviations from procedure.
5. Providing experience feedback from other programs.

Pre-launch Checkout

The largest and most sophisticated part of space vehicle testing, pre-launch checkout, has been intentionally omitted from specific mention in this discussion. This was done because its basic purpose is to confirm that all prior assembly and test has been properly done and its scope is so great that adequate coverage would require a separate paper.

Concluding Comments

Two of the most important phases of the Apollo program in terms of contribution to its success are testing of hardware and training of people. There are many aspects of planning, discipline, and motivation as used in this program which could be used constructively in other types of technological development programs. With proper consideration of economic factors and tailoring of the techniques to meet the need, some of these procedures could be used as well in transportation, construction, and consumer products manufacturing.

Fatigue Life Prediction for Weldments with Internal Cavities

REFERENCE: van der Zanden, A. M., Robins, D. B., Topper, T. H., "Fatigue Life Prediction for Weldments with Internal Cavities," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 268-284.

ABSTRACT: A method evolved for determining the local stress-strain response of notched members is extended to predict fatigue resistance of weldments with internal cavities. It is applied to both stress-relieved samples and as-received weldments. Test results from stress-relieved samples show good agreement with life predictions based on prestrained fully reversed strain controlled smooth specimen tests.

A calculation of inelastic deformations induced in the weld metal of as-received samples indicates that localized stresses will be of yield stress magnitude on the first load application. A fatigue parameter then incorporates ensuing mean stress effects. Good life predictions obtained in this application, where the local stress history is complex, demonstrate the adequacy of this method of analysis.

KEY WORDS: predictions, fatigue (materials), failure, weldments, stresses, crack initiation, plastic deformation, strains, deformation, stress concentration, residual stress

Nomenclature

Δs	Nominal stress range
Δe	Nominal strain range
$\Delta \sigma$	Local stress range
$\Delta \epsilon$	Local strain range
$\Delta \epsilon_p$	Plastic strain range
σ_a	Local stress amplitude
σ_m	Maximum stress
ϵ_a	Local strain amplitude
s_a	Nominal stress amplitude

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¹ Present and former research assistants, and professor, respectively, Department of Civil Engineering, University of Waterloo, Waterloo, Ontario, Can.

- e_a Nominal strain amplitude
- σ_f' Fatigue strength coefficient, value of true stress given by the linear log σ_a versus log $2N_f$ plot at $2N_f = 1$
- ϵ_f' Fatigue ductility coefficient, value of true plastic strain given by the linear log $\Delta\epsilon_p/2$ versus log $2N_f$ plot at $2N_f = 1$
- K_f Fatigue concentration factor
- K_t Theoretical stress concentration factor
- K_σ Stress concentration factor
- K_ϵ Strain concentration factor
- a Material constant
- b Fatigue strength exponent, slope of the linear log σ_a versus log $2N_f$ plot
- c Fatigue ductility exponent, slope of the linear log $\Delta\epsilon_p/2$ versus log $2N_f$ plot
- r Notch root radius
- E Elastic modulus
- $2N_f$ Reversals to failure
- n' Cyclic strain hardening exponent, slope of the linear log $\Delta\epsilon_p/2$ versus σ_a plot.

Fatigue failures in weldments initiate at external changes in geometry or at internal flaws. In both cases, fatigue resistance is determined by the behavior of the metal at these highly stressed regions where fatigue cracks initiate. As with other notched components, localized plastic deformations occur infrequently although service stresses and strains are usually nominally elastic. In weldments, moreover, this local plasticity is experienced on the first load application due to residual stresses induced during the thermal cycle of the welding process. Life prediction based on the stress-strain history at a notch root is of recent development[1-4]. Most of the difficulties previously experienced in making accurate life predictions were due to analytical methods being based on nominal rather than local response. An analysis similar to the one employed for notched components[1] will be used to predict the fatigue response of weldments with flaws. It may be noted that most of the weldments tested contained flaws in the form of gas cavities emanating from the region of incomplete weld penetration.

Analysis

In general the method of fatigue analysis employed in this paper consists of first determining the local stress-strain response at the crack initiation region adjacent to weld cavities in terms of the applied loads. Fatigue predictions are then based on this local stress-strain history suitably accounting for prestrains, initial residual stresses, and ensuing mean stress levels.

Local Stress-Strain Response

This analysis employs a relation between nominal and local stresses and strains suggested by Neuber[5]

$$K_t^2 = K_\sigma \cdot K_\epsilon \quad (1)$$

where

K_σ = stress concentration factor,

K_ϵ = strain concentration factor,

K_t = theoretical stress concentration factor.

This equation has been modified in order to determine stress-strain response at the notch root under cyclic conditions[1]

$$\Delta\sigma \cdot \Delta\epsilon = K_f^2 \cdot \Delta s \cdot \Delta e \quad (2)$$

where

Δs = nominal stress (net section) range,

Δe = nominal strain range,

$\Delta\sigma$ = local stress range,

$\Delta\epsilon$ = local strain range,

K_f = fatigue concentration factor.

Local response is obtained from simultaneous solution of Eq 2 and a cyclic stress-strain equation in the form suggested by Morrow[6] namely,

$$\frac{\Delta\epsilon}{2} = \frac{\Delta\sigma}{2E} + \epsilon_f' \left(\frac{\Delta\sigma}{2\sigma_f'} \right)^{\frac{1}{n'}} \quad (3)$$

where

σ_f' = fatigue strength coefficient,

ϵ_f' = fatigue ductility coefficient,

E = elastic modulus,

n' = cyclic strain hardening exponent,

$\Delta\sigma$ = local stress ranges,

$\Delta\epsilon$ = local strain ranges.

Since Eq 2 deals with ranges of stress and strain, its solution must be on a reversal by reversal basis.

The accuracy of Eq 2 will be deliberately tested on internal cavities found in weldments where deformation more closely approximates plane strain than plane stress. In defence of this application, it should be noted that inspection of Eq 1 derived for shear-strained notches[5], verified for notches in plane stress[7] and extended to notches under plane stress cyclic conditions[1], indicates that errors in plane strain applications should not be large.

Note that K_t in Eq 1 has been replaced by K_f in Eq 2 to account for the reduced effect of stress concentration in fatigue. A review[8] of

several relationships, between K_f and K_t , found none to be markedly superior; but concluded that Peterson's relation

$$K_f = 1 + \frac{K_t - 1}{1 + a/r} \quad (4)$$

where r is the notch radius and a is the material constant, was conservative and reasonably accurate. Thus, Eq 4 will be chosen as illustrative of the present level of treatment for notch reduction factors.

Life Prediction

Life is now predicted utilizing a relationship suggested by Morrow[6] that is reasonably accurate for most engineering materials,

$$\epsilon_a = \frac{\sigma_f'}{E} \cdot (2N_f)^b + \epsilon_f' \cdot (2N_f)^c \quad (5)$$

where

b = fatigue fracture exponent,

c = fatigue ductility exponent,

$2N_f$ = number of reversals to failure,

ϵ_a = local strain amplitude.

Making use of the Coffin-Manson[9,10] and Basquin[11] relationships used in deriving Eq 5 we can derive a fatigue-life relation of the form

$$\sigma_a \cdot \epsilon_a = \frac{(\sigma_f')^2}{E} \cdot (2N_f)^{2b} + \sigma_f' \cdot \epsilon_f' \cdot (2N_f)^{b+c} \quad (6)$$

where σ_a is the local stress amplitude. Smith et al[12] developed a similar relation

$$\sigma_m \cdot \epsilon_a = \frac{(\sigma_f')^2}{E} \cdot (2N_f)^{2b} + \sigma_f' \cdot \epsilon_f' \cdot (2N_f)^{b+c} \quad (7)$$

where the maximum stress σ_m has replaced the stress amplitude σ_a in Eq 6. Note that Eqs 6 and 7 are identical for zero mean stresses. The unique feature of Eq 7 which may be obtained from fully reversed strain controlled smooth specimen tests is that it is capable of predicting life of specimens subjected to various mean stresses[12].

Life reduction caused by prestraining has been shown to be pronounced when plastic deformation is followed by predominantly elastic strain levels[13-15]. A fatigue curve based on prestrained specimens will in general give more conservative life predictions than the curve obtained from nonoverstrained tests[13]. Since the high levels of residual stresses in as-received weldments result in plastic straining at internal cavities, at least on the first load application, all life predictions are based on data obtained from tests of prestrained smooth weld metal specimens.

Material and Specimens

Specimens used in this investigation were machined from welded wide flange beams (WWF) in which welding consisted of a submerged arc process conforming with specifications AWS-5.17-65 and AWS-F60. Physical properties and chemical compositions of the as-deposited weld metal and parent metal (CSA G40.12) a semikilled general purpose structural steel are listed in Table 1.

TABLE 1—*Physical properties and chemical composition of parent metal (CSA G40.12 structural steel) and as-deposited weld metal obtained from welded wide flange beams.*

<i>Physical Properties</i>	<i>Parent Metal</i>	<i>Weld Metal</i>
Elastic Modulus, ksi	29×10^3	29×10^3
Lower Yield Stress, ksi	55	85
True Fracture Strength, ksi	161	137
True Fracture Strain	1.00	0.60
BHN-Hardness	143	175
<i>Chemical Composition^a</i>		
<i>Element</i>		
Carbon	0.187	0.160
Manganese	1.29	1.14
Phosphorous	0.010	0.009
Sulphur	0.019	0.016
Silicon	0.04	0.18
Iron	balance	balance

^a Check analysis—percent by weight.

1. Smooth parent metal specimens were machined from the flange plate of a WWF beam to the same dimensions as the smooth weld metal specimens in Fig. 1(a).

2. Smooth weld metal specimens were machined from WWF beams keeping the test section in the weld metal as shown in Fig. 2(a).

3. Notched weld metal specimens were made by drilling a 0.025-in. diameter hole centrally located and transverse to the smooth weld metal specimen's axis.

4. Mechanically stress-relieved samples hereafter referred to as parent-weld specimens were also machined from the welded joint as shown in Fig. 2(b). A photograph of this specimen appears in Fig. 1(b).

5. As-received samples, hereafter referred to as T specimens, were cut from WWF beams. Figure 3(a) is a photograph of this specimen.

Test Program

The test program designed to develop and evaluate a life prediction method for weldments containing cavities is outlined below. Testing was done in a servo controlled, closed loop electrohydraulic machine[16].

1. Smooth parent metal specimens were subjected to constant amplitude, strain controlled cycling to failure.

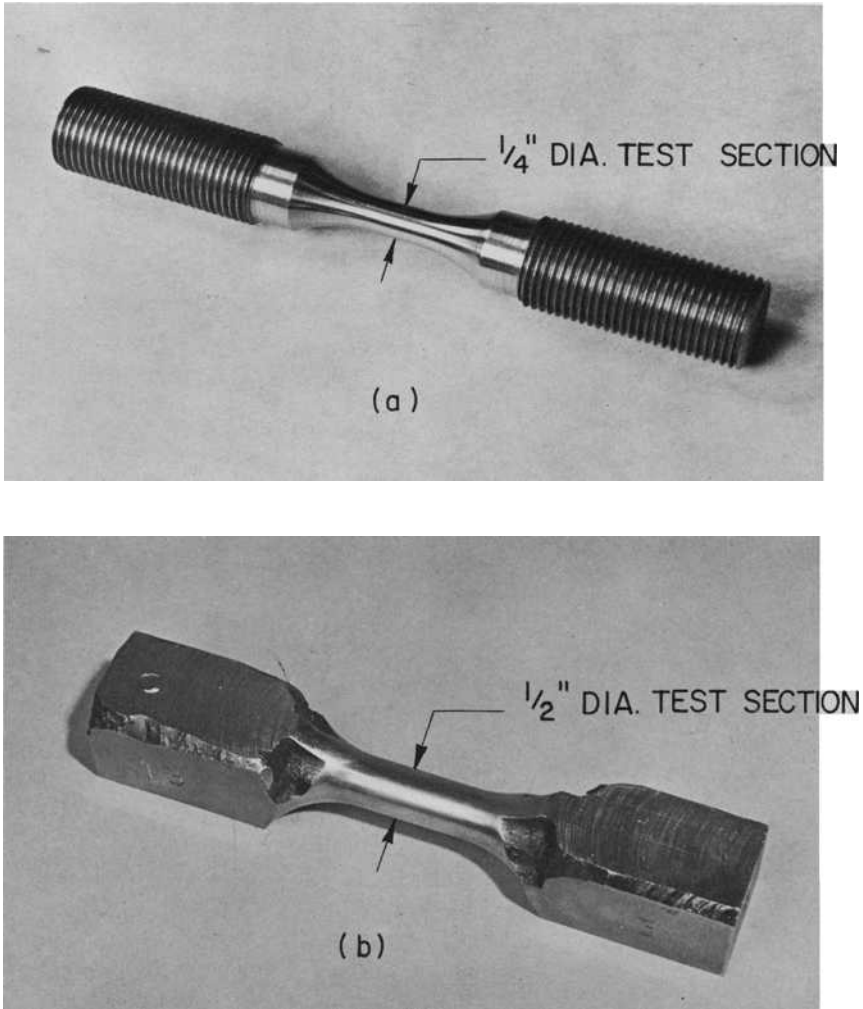


FIG. 1—(a) Smooth weld metal specimen machined from WWF beam; (b) parent-weld metal specimen machined from a WWF beam.

2. Smooth weld metal specimens in deformation control were subjected to an overstraining sequence (described later), which gave peak strain amplitudes of 0.01. This sequence was then followed by constant amplitude cycling to failure.

3. Notched weld metal specimens in edge strain control received the same overstraining sequence as the smooth weld metal specimens. The effect of high notch root peak strains, somewhat greater than 0.01, is not expected to be significant. However, this effect is incorporated in life predictions. These specimens were then cycled to failure in fully reversed constant amplitude strain control.

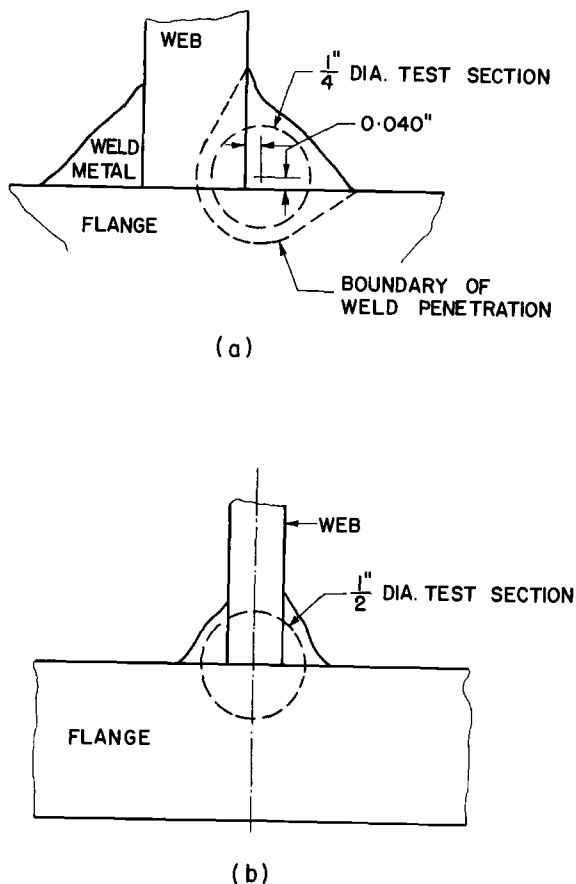
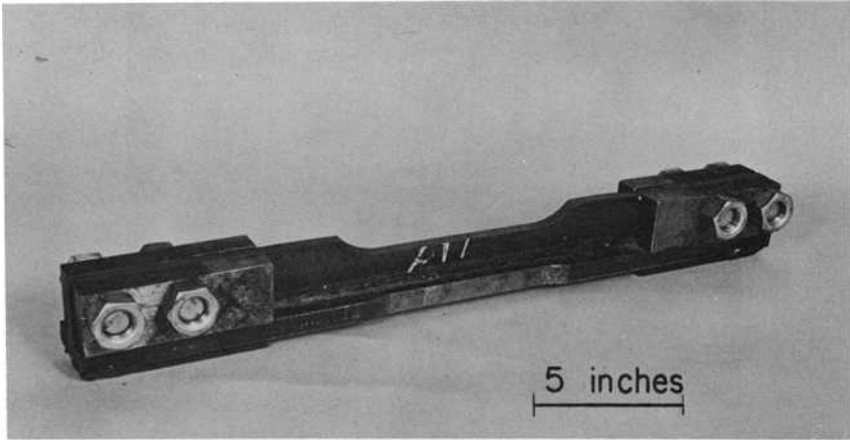


FIG. 2—(a) Location of $\frac{1}{4}$ -in. test section of notched and smooth weld specimens; (b) location of $\frac{1}{2}$ -in. test section of parent-weld specimens.

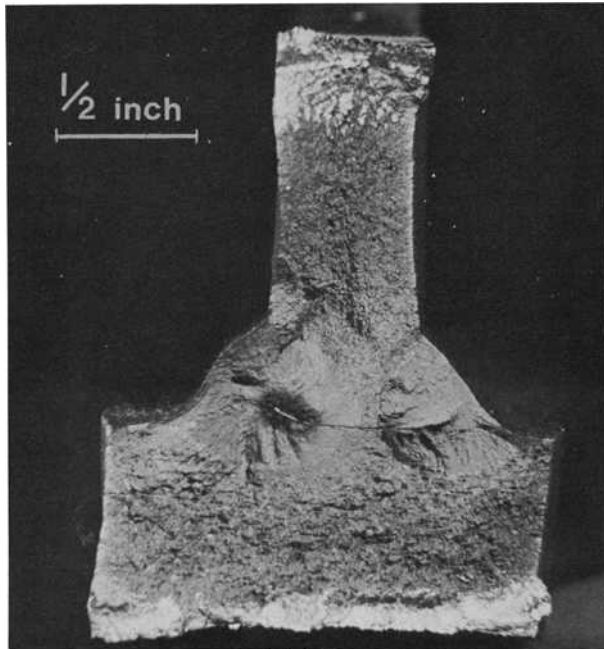
4. Parent-weld specimens like the notched weld metal specimens received an overstraining sequence followed by fully reversed strain cycling to failure.

5. T specimens were simply subjected to constant amplitude (O-max) load controlled cycling to failure.

The overstraining sequence applied to all specimens except smooth parent metal and T specimens, was designed to plastically prestrain the samples and leave them free of residual stresses. Figure 4 demonstrates how an induced residual stress in a smooth weld metal specimen is relieved by this overstraining sequence. An initially applied mean strain induces a tensile stress equivalent to the residual stress present in as-received weldments. The subsequent overstraining sequence consisting of 21 variable amplitude cycles (see Fig. 4) reduces this residual stress to zero.



(a)



(b)

FIG. 3—(a) *T* specimen cut from a WWF beam; (b) fracture surface of a *T* specimen.

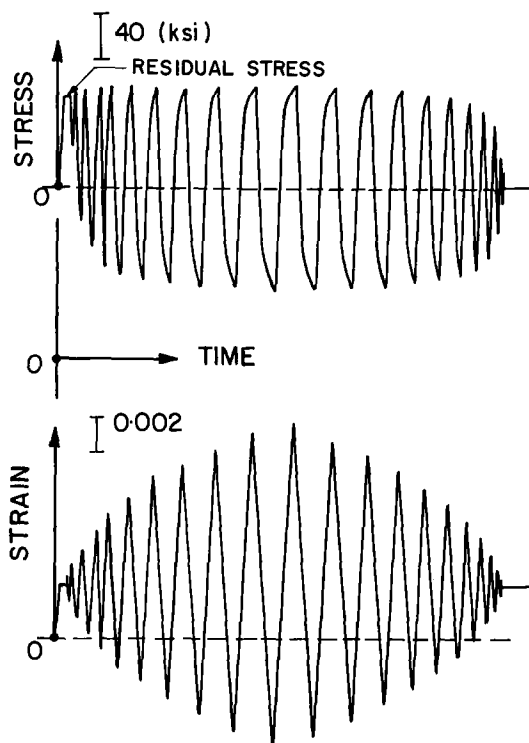


FIG. 4—Strain controlled overstraining sequence showing relaxation of mean stress.

Failure in all tests is defined as separation of the specimen into two parts, namely fracture.

Results

Smooth Specimens

For all strain amplitudes examined, cyclic stress response for smooth specimens of both parent metal and weld metal stabilized within the first ten percent of life and thereafter remained nearly constant until a crack appeared. These stabilized stress amplitudes and corresponding strain amplitudes are used to obtain the cyclic stress-strain relations given by Eq 3. Cyclic test data for the parent and weld metals appear in Fig. 5.

Weld metal fatigue data are plotted in Figs. 6 and 7 on the basis of strain amplitude (elastic, plastic and total) and the product of stress and total strain amplitude, respectively, against number of reversals to failure. Both Eqs 5 and 6 (constants b , c , σ_f' and ϵ_f' of the Coffin-Manson[9,10] and Basquin[11] relationships are derived from a least-squares fit to elastic and plastic strain amplitude versus reversals to failure data) adequately represent fatigue behavior of the weld metal.

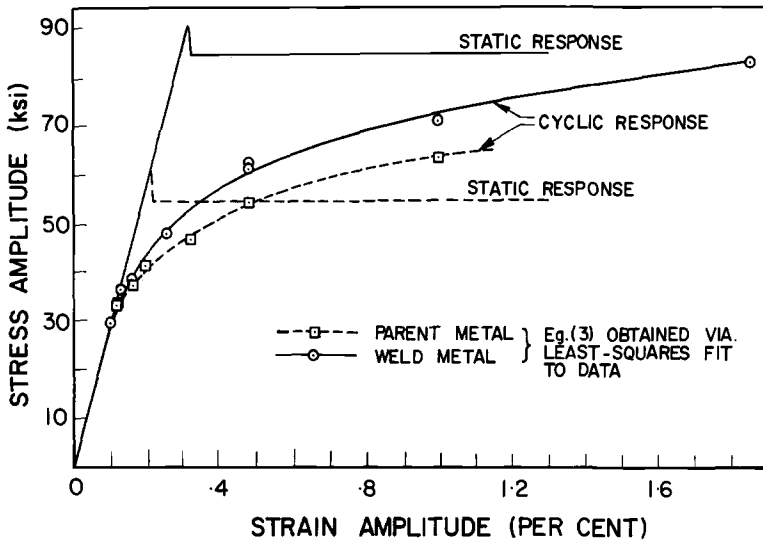


FIG. 5—Static and cyclic stress-strain response of parent and weld metals.

Notched Weld Metal Specimens

The material “size effect” constant a in Eq 4 is obtained from a comparison of smooth and notched weld metal specimen tests. Data from these tests are plotted on coordinates of $s_a \cdot e_a$ versus life in Fig. 8, where s_a and e_a are the nominal stress and nominal strain amplitudes. By assuming that identical values of the parameter $\sigma_a \cdot \epsilon_a$ for both smooth and

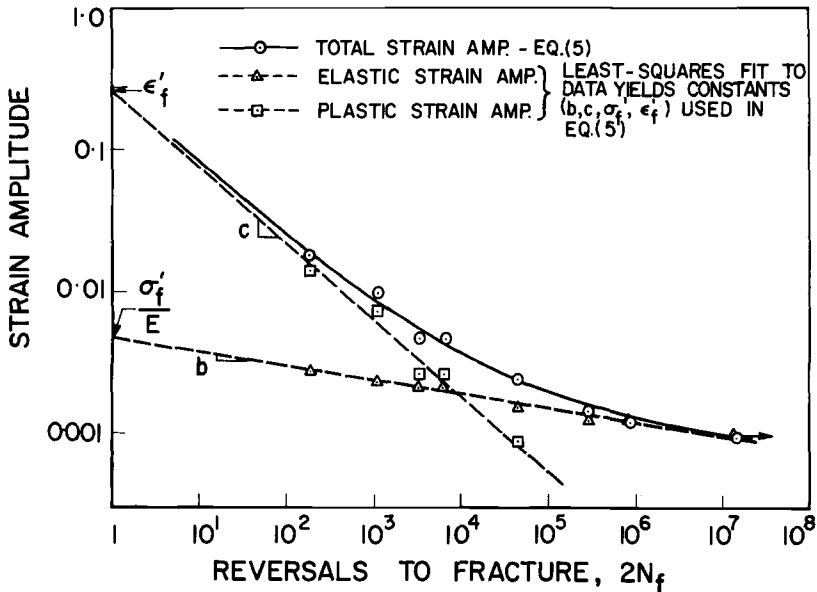


FIG. 6—Fatigue of prestrained weld metal under fully reversed strain cycling.

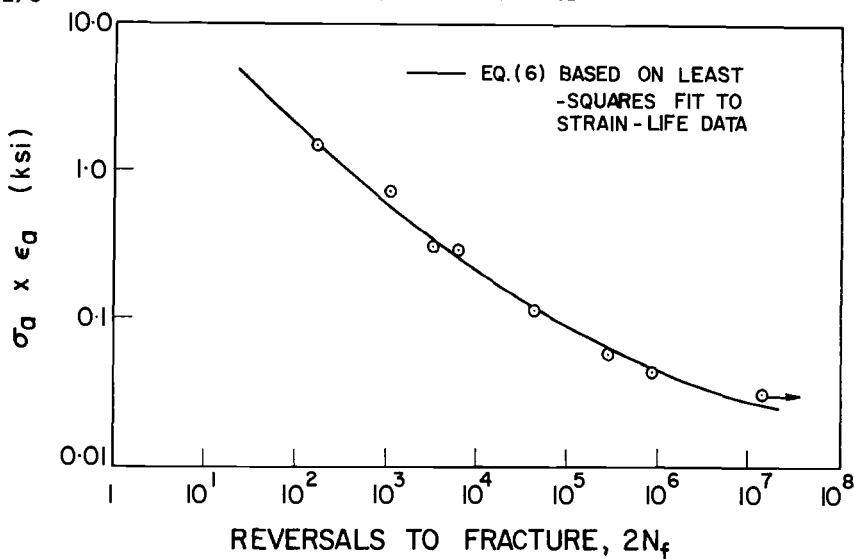


FIG. 7—Fatigue in terms of the product of stress and strain for prestrained weld metal under fully reversed strain cycling.

notched weld metal specimens give identical lives then the vertical distance between the curves in Fig. 8 is a measure of K_f^2 . Agreement between notched weld metal data points and the curve drawn through them is good but some decrease in K_f^2 is observed at intermediate lives (10^4 to 10^6 reversals). Utilizing the experimentally determined value of $K_f^2 = 2.0$ and the theoretical stress concentration factor $K_t = 2.89$ computed from Ref 17, Eq 4 gives $a = 0.045$.

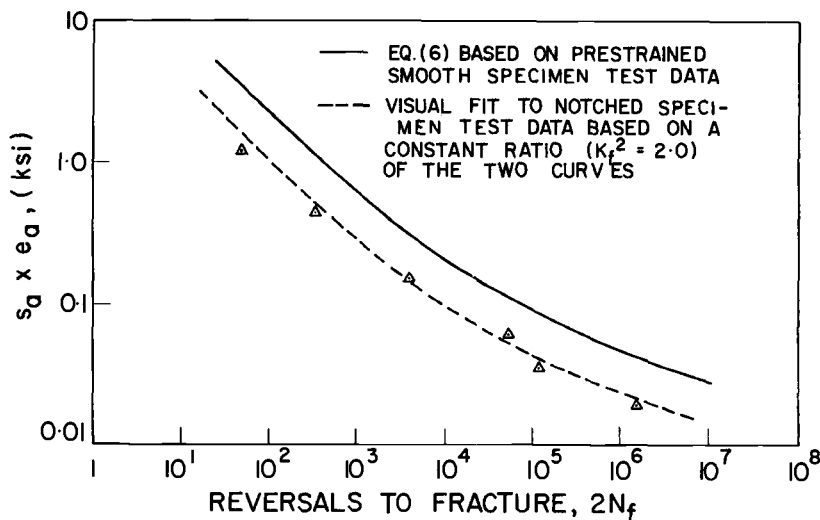


FIG. 8—Comparison of notched weld metal specimen test data with the smooth specimen fatigue prediction curve.

Parent-Weld Specimens

As discussed in the preceding section, parent-weld specimens machined from the weldment were overstrained to remove residual stresses and then cycled to failure under fully reversed constant amplitude strain control. Values of $K_f^2 \cdot s_a \cdot e_a$ for each test are compared with the equivalent smooth specimen curve represented by Eq 6 in Fig. 9. This comparison shows predictions to be accurate though conservative in the middle life region where a similar conservative trend was noted for notched weld metal samples (see Fig. 8).

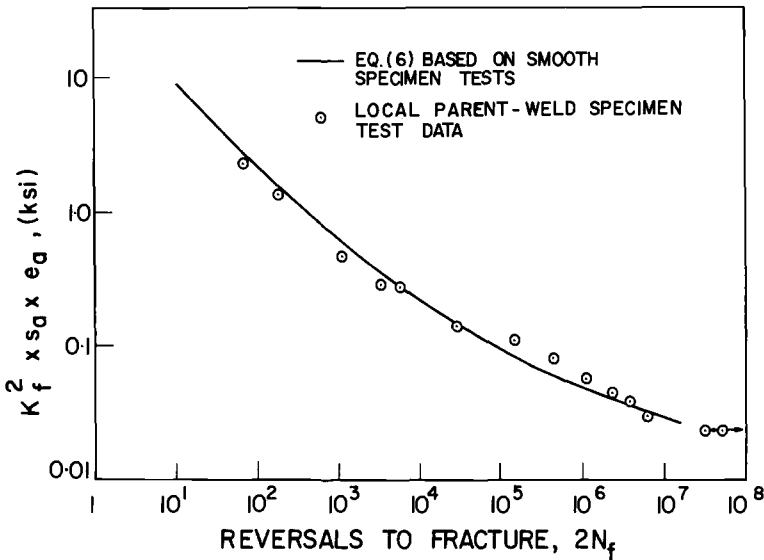


FIG. 9—Fatigue prediction curve for fully reversed strain controlled tests of prestrained parent-weld specimens.

Fatigue concentration factors (K_f) used with parent-weld specimens and T specimens are calculated from Eq 4. Theoretical stress concentration factors (K_t) in Eq 4 are obtained from Ref 18 based on dimensions of the cavities revealed by fracture surfaces. However, in service applications, dimensions of cavities in welded components can be obtained from nondestructive techniques such as X-ray diffraction[19]. For those cases where more than one cavity appears on the fracture surface (for example, Fig. 3(b) shows two cavities), the one associated with the largest value of K_f is assumed to cause failure. It may be noted that the cavity projecting to the left from the region of incomplete penetration in Fig. 3(b) is a typical one measuring approximately 0.015 in. in diameter by 0.10 in. long.

T Specimens

The calculation of local stresses and strains in T specimens is complicated by the welding process. During the cooling portion of the thermal cycle induced by the welding process, the metal in the weld zone attempts to shrink but is restrained by surrounding cooler metal. Large inelastic strains result, leaving tensile residual stresses in the weld zone that have been shown by theoretical[20] and experimental studies[21] to be of yield stress magnitude. Some relief of this residual stress may, however, occur when the T specimen is cut from the I beam.

Although calculations of stress and strain values in the weld metal due to deformations accompanying cooling[22] are approximate, little error in ensuing mean stress will result because the flatness of the material stress-strain curve beyond the yield point makes the peak stress insensitive to strain. Figure 10 shows a sample local stress-strain calculation. The monotonic curve is obtained by assuming an elastic-perfectly-plastic response during first loading to reach the hyperbola, Eq 2, (labelled 1). Peak stress and strain values for this loading are given by the intersection of the monotonic stress-strain curve with this hyperbola. On reversal of loading direction, a new hyperbola (labelled 2) with a value of $K_f^2 \cdot \Delta s \cdot \Delta e$ determined by the second nominal stress range, is constructed with its origin at the previous peak. On this second, and subsequent reversals, the cyclic stress-strain curve is used. This process is repeated reversal by reversal, to determine the local stress-strain history.

Mean stress relaxation was estimated from previously determined relaxation data of smooth specimens to arrive at an average value of the fatigue parameter $\sigma_m \cdot \epsilon_a$ for each test. Some increase in accuracy could have been achieved by simulating the notch root stress-strain response using Eq 2, but considering the approximations involved in determining residual stresses, this procedure was not warranted.

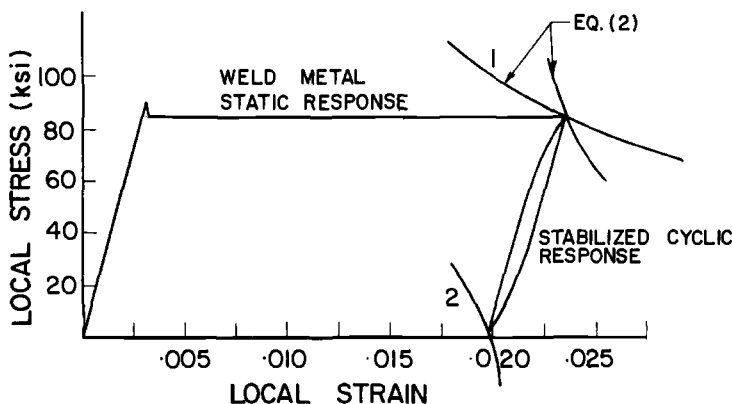


FIG. 10—Determination of local stress-strain response for O-max load cycling of T specimens.

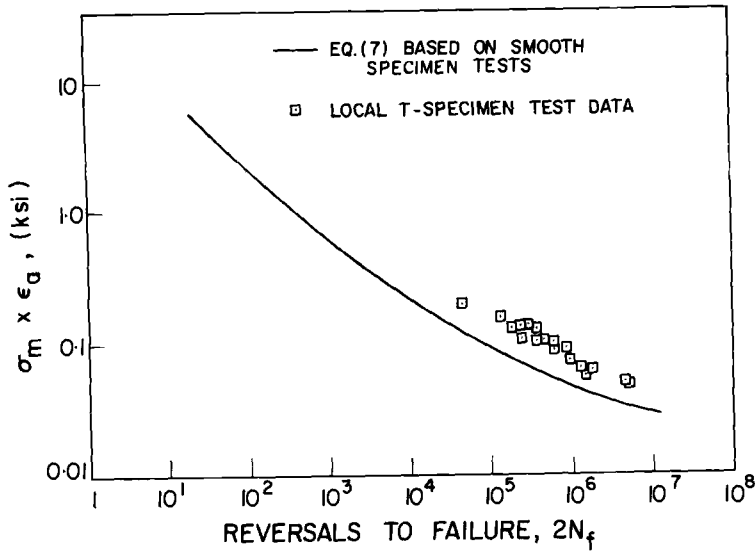


FIG. 11—Fatigue prediction curve for *O*-max stress controlled cycling of *T* specimens.

The values of $\sigma_m \cdot \epsilon_a$ for the *T* specimens are compared with the equivalent smooth specimen curve represented by Eq 7 in Fig. 11. A trend similar to that found in Figs. 8 and 9 is again observed with the predicted values being somewhat conservative in the middle life region. Deviations from predictions are even greater for these specimens than for the notched weld metal and parent-weld metal specimens.

Discussion

Notched specimens of weld metal showed good agreement with the method of analysis developed from Eq 2, but K_f decreased in the middle life region. Parent-weld specimens with internal cavities, again, had somewhat greater fatigue resistance than that given by this analysis in the same life region. A slightly more pronounced trend was observed for *T* specimens with initial residual stresses and subsequent mean stresses. Since this trend in fatigue resistance is similar in notched weld metal specimens, stress relieved parent-weld specimens and *T* specimens, it does not seem to be due to residual or mean stresses.

This increased discrepancy in fatigue prediction with increasing specimen size may be due to an increasing crack propagation stage since the method of analysis used is accurate only to relatively short crack lengths. Additional errors are no doubt involved in extending Eq 2 to plane strain conditions near internal cavities and in estimating residual stresses in the *T* specimens. Predictions of fatigue lives for specimens with in-

ternal cavities is, however, accurate enough in both cases to recommend this method for design applications.

Conclusions

1. Extension of the local stress-strain approach of Ref 1 to internal weld cavities gives conservative life predictions for mechanically stress relieved and as-received welded samples in the middle life region (10^4 to 10^6 reversals).

2. The analysis based on local stress-strain response at stress raisers such as cavities is adequate for predicting fatigue behavior of weldments.

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DISCUSSION

R. S. Carey¹—Point of clarification on method of determining K_t (K_f) of stress concentration caused by incomplete penetration of T joint welded specimens stressed along axis of T joint.

A. M. van der Zanden, et al (authors)—Incomplete weld penetration invariably causes expansion of trapped air during the welding process resulting in the formation of gas cavities. The shape of these cavities is then approximated by some known geometry such as an ellipsoid for which K_t is available in the literature. In most industrial applications the geometry of the critical location will not be known prior to fracture. Nondestructive techniques can be used to determine geometries of these critical locations. If these techniques are not readily available then from our experience $K_t = 3.0$ and $K_f = 1.5$ are typical values for the case of incomplete weld penetration.

C. C. Yen²—How can we use your prediction method as a fatigue cumulation rule?

A. M. van der Zanden, et al—Essentially the method predicts the local stress-strain response from a known nominal loading history. A digital computer program would facilitate localized damage computation, reversal by reversal, based on a simple linear damage rule such as the one proposed by Miner.³

This damage accumulation procedure represents a first order approximation to material response since it does not include such cycle dependent behaviour as hardening, softening, or mean stress relaxation. However, since the weld metal quickly stabilizes with cycling, little error should be introduced by ignoring hardening and softening behaviour. Also, by ignoring mean stress relaxation damage predictions will be con-

¹ Chief Metallurgical Engineer, Spinco Division, Beckman Instruments, Inc., Palo Alto, Calif.

² Member of the Technical Staff, Fatigue and Fracture Mechanics Group, Department of Engineering, Los Angeles Division, North American Rockwell Corp., Los Angeles, Calif.

³ Miner, M. A., *Journal of Applied Mechanics*, JAMCA, Series E, Transactions, American Society of Mechanical Engineers, Vol. 12, 1945, p. A-159.

servative. Existing damage models^{4,5} include most of this transient behaviour but in view of the simplifying assumptions, this model should still yield reasonably accurate results.

This work is in progress at the University of Waterloo.

⁴ Martin, J. F., Topper, T. H., and Sinclair, G. M., *Materials Research and Standards*, MTRSA, Vol. 11, No. 2, Feb. 1971.

⁵ Jhansale, H. R. and Topper, T. H., "Cyclic Deformation and Fatigue Behavior of Axial and Flexural Members—A Method of Simulation and Correlation," Proceedings of the First International Conference on "Structural Mechanics and Reactor Technology," Berlin, Vol. 5, L 5/3, 20-24, Sept. 1971.

Predictive Testing of Aircraft Structures

REFERENCE: Rosenfeld, M. S., "Predictive Testing of Aircraft Structures," *Testing for Prediction of Material Performance in Structures and Components, ASTM STP 515*, American Society for Testing and Materials, 1972, pp. 285–313.

ABSTRACT: The historic development of airframe structural testing is reviewed and the role played by structural testing in the development of reliable airframe structures is discussed. The threefold safety-factor, safe-life, and fail-safe design approach provides reliable airframe structures when complemented by an effective laboratory test program that includes static, dynamic, and fatigue tests.

Using an iterative analytical and test approach, the operational strength limits of an airframe can be predicted fairly accurately from the static test results. However, predictions of the safe service life from fatigue test data alone leaves much to be desired because of the large variations in service usage. Nevertheless, the alarming cost of modifying aircraft in service has forced an almost complete acceptance of laboratory structural fatigue testing as an economical way of doing business.

The ability to predict the lives (actually the lives remaining) of individual aircraft depends upon the development of an instrument that would measure the fatigue damage actually incurred. Until that time, frequent and rigidly controlled periodic inspections must be performed on all aircraft, and an elaborate scheme of progressive preventive maintenance is mandatory.

KEY WORDS: predictions, simulation, service life, fatigue tests, static tests, airframes, dynamic tests, failure, plastic deformation, stresses, strains, yield strength

Frequently the necessity for the extensive structural test program required under all military aircraft contracts is questioned by those who feel that present analytical methods are sufficiently accurate to predict the structural behavior, and that money thus saved can be used either to reduce the overall contract cost or to obtain more aircraft for the same overall cost. The speciousness of the arguments against structural testing are demonstrated by the increasing reliance on structural testing by commercial aircraft manufacturers even though contract and FAA requirements do not specify such testing.

The opinions or assertions expressed in this paper are the private ones of the author and are not to be construed as official or reflecting the views of the Department of the Navy or of the Naval service at large.

¹ Aero Structures Dept, Naval Air Development Center, Warminster, Pa. 18974.

This paper is devoted to answering the question "Why do we do structural testing?" There are many answers to this question but the underlying purpose of these tests is to ensure the integrity and reliability of the aircraft structure for the intended mission(s) and for the desired life with a minimum of maintenance. This is a rather tall order for minimum weight structures with the low factor of safety necessary for high payload, high performance aircraft. The specific reasons for the various tests will be covered as each type of test is discussed, along with the application of the test data toward prediction of the service structural behavior.

Static Tests

For many years, static testing of aircraft structures was the only type of testing performed, other than flight testing. In the early days static testing was performed by simply adding dead weights to the test structure in a predetermined manner until the structure collapsed. Tests were performed primarily on wings, fuselage, and tail surfaces since they had large expanses over which dead weights could be added.

As the loads increased in magnitude and complexity, simple hydraulic systems were used to apply loads to discrete hard points of the structure. Tension pad and lever systems were developed for load application to simulate the effect of distributed aerodynamic forces. These systems are still in use today except for the addition of closed loop servo-controlled hydraulic systems.

The objectives of the static test program are: (1) to ensure that the basic design is structurally adequate for the required ultimate design loads; (2) to obtain strain data for verification of the stress analysis; (3) to determine the growth potential available in the structure; and (4) to alleviate and prevent (where possible) future structural maintenance difficulties.

For military aircraft, at least one airframe of each model is static tested in accordance with specifications to the ultimate loads for all critical conditions. For Naval aircraft, each major structural assembly is tested to failure for what is considered to be its most critical conditions; Air Force requirements are similar. Tables 1 and 2 which are taken from the military test specification (MIL-A-8867) outline the static tests required for wing and fuselage structures. There are four more pages similar to these outlining the static test program for the other major structural assemblies and an additional half page of special Navy requirements for carrier based aircraft. For each test the distribution of applied loads must simulate the actual distribution as closely as is practicable, the loads resulting from internal pressurization must be simulated, and temperature effects from internal heat sources and from aerodynamic heating which may produce both significant strength reduc-

TABLE 1—*Wing group static tests.*

Test	Loading condition	Magnitude of load	Special requirements
Wing	Symmetrical positive low angle of attack, in more critical of basic and high-drag configurations	Failing	Measure aileron displacements, and in closed aileron control systems measure loads resulting from wing deflection under limit load
Wing	Symmetrical positive high angle of attack, in more critical of basic and high-drag configurations	Design ultimate	
Wing	Rolling pull-out—in more critical of basic and high-drag configurations in maximum load factor condition if maximum and minimum load factor conditions are approximately equally critical	Design ultimate	
Wing	Auxiliary float load critical for wing	Design ultimate	
Flap and supporting structure	Critical	Failing	
Aileron and supporting structure	Critical	Failing	
Aileron tab and supporting structure	Critical	Failing	Test each type of tab
Wing-folding mechanism and supporting structure	Critical	Design ultimate	
Slat, spoiler, wing-mounted speed-reduction device, leading edge droop, and supporting structure	Critical	Failing	Test each type of slat, spoiler, wing-mounted speed-reduction device, leading edge droop, and supporting structure

tions and significant induced stresses must be simulated or compensated for in appropriate tests.

To accomplish such a test program economically, we simply support the airframe at two points, construct a large test framework around the airframe, install the load distribution system, hydraulic actuators and load cells between the airframe and the test framework, install the simu-

TABLE 2—*Fuselage static tests.*

Test	Loading condition	Magnitude of load	Special Requirements
Fuselage	Horizontal tail load—in condition that is critical for fuselage	Fuselage failing	Test only to design—ultimate load if the tests specified below are more critical
Fuselage	Vertical tail load—in condition that is critical for fuselage	Fuselage failing	Test only to design—ultimate load if the tests specified above or below are more critical
Fuselage	Tail load—in condition that is critical for fuselage in torsion	Fuselage failing	Test only to design—ultimate load if the tests specified above are more critical
Fuselage	Flight load—in condition that is critical for fuselage in way of and forward of the wing	Fuselage failing	Test only to design—ultimate load if tests for landing conditions are more critical
Fuselage mounted speed-reduction device and supporting structure	Critical	Failing	Test each type of fuselage mounted speed-reduction device
Fuselage	Auxiliary landing gear—unsymmetrical load condition critical for fuselage	Design ultimate	

lated aerodynamic heating system, counterbalance the airframe and the heating and loading systems, and add strain and deflection measuring systems. The resulting setup for a combined wing and fuselage test with aerodynamic heating is shown in Fig. 1. Of course, not all setups are this complicated. Figure 2 shows a much simpler setup for the wing test of an air-sea rescue aircraft. When it is desirable to expedite the test program, a major structural assembly may be tested in a jig as shown in Fig. 3.

After performing a number of static tests, what have we learned? First, we have determined the actual strength of each structure, not as designed, but as built. Shortly after World War II a survey was made to determine static strength experience; the results are depicted in Fig. 4. Obviously our stress analyses were neither sufficiently accurate nor comprehensive to predict the initial failures depicted; however, not all of the failures resulted from inadequate analysis, some could be blamed on our manufacturing and inspection methods. Nevertheless, our analytical prediction methods left much to be desired, and after redesign and retest, all structures can equal or exceed the appropriate design ultimate load; that is, they have positive margins of safety.

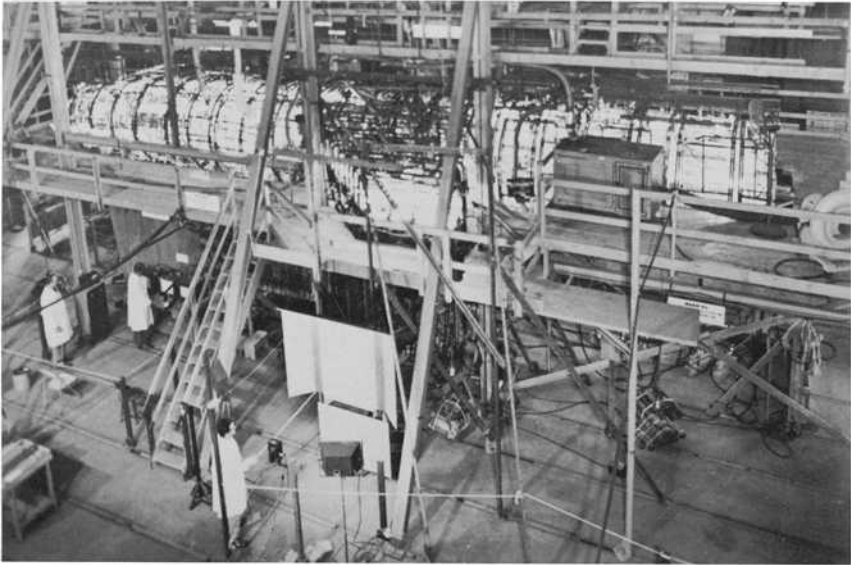


FIG. 1—Elevated temperature static test setup.

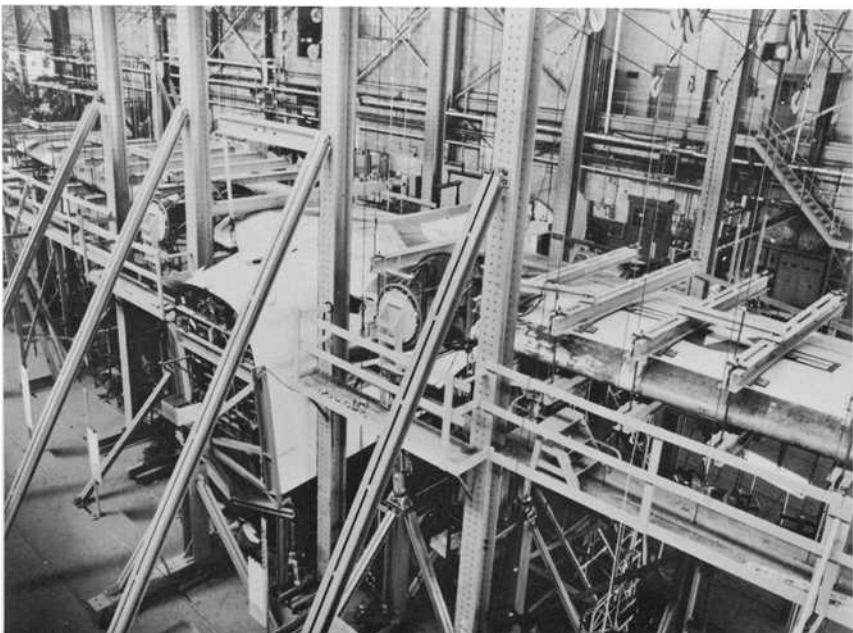


FIG. 2—Wing test setup for air-sea rescue airplane.

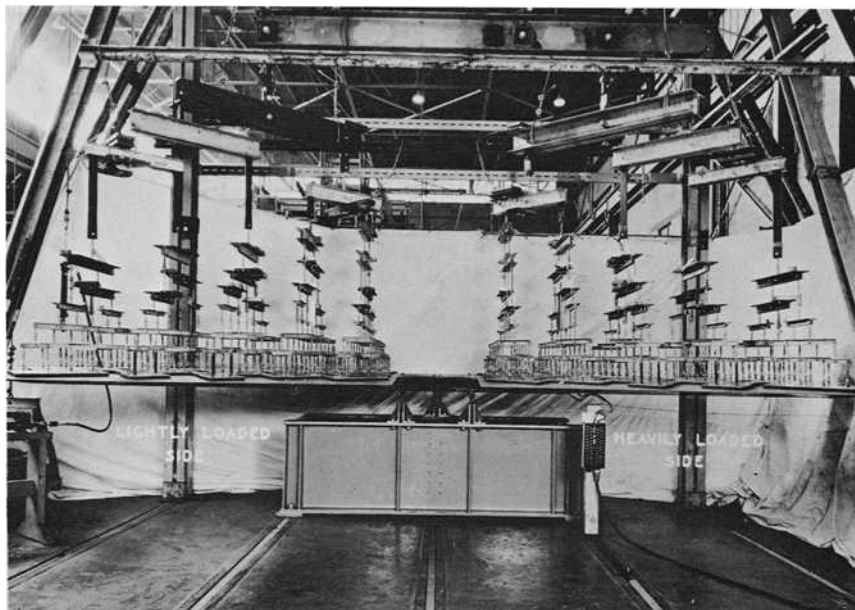


FIG. 3—Horizontal tail test setup.

Now that we have an airplane with sufficient strength (and analytically predicted adequate life), we deliver these aircraft to operating squadrons. Initially the aircraft are used as intended, but soon the pilots discover that the performance and payload capabilities exceed the restrictions thus establishing an "operational requirement" for increasing the payload or changing the mission or both. The static test results are used to predict the feasibility of these "operationally required" changes, and new restrictions are established for the heavier payload conditions that reduce

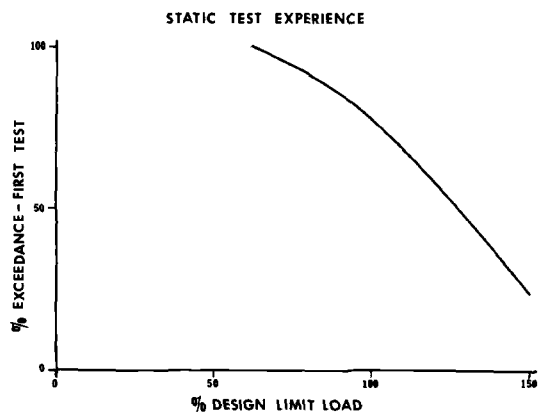


FIG. 4—Static test experience.



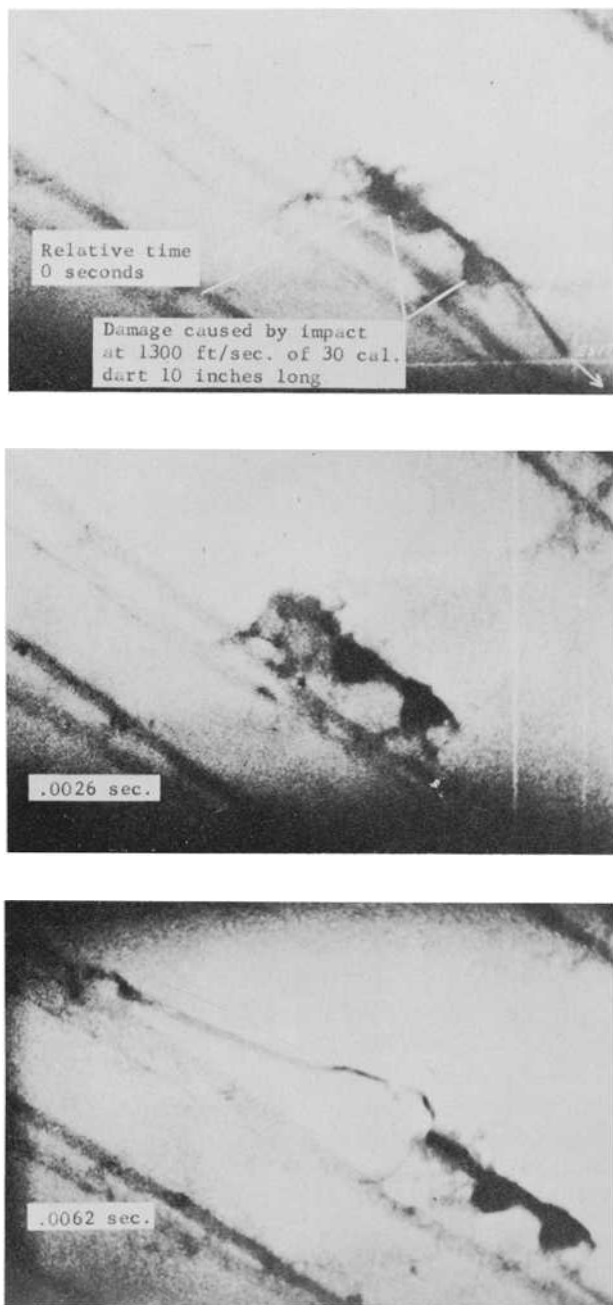
FIG. 5—*Light attack aircraft.*

the apparent margins of safety to zero. The aircraft then looks like the one in Fig. 5.

Lately, static testing has taken on a new role—the determination of residual strength after a given amount of structural damage has occurred. In some tests a major structural part is severed prior to loading and the structure is loaded to 100 percent limit load. If the structure does not fail, the damage is increased until failure occurs at this load. This type of testing is used to determine maximum tolerable structural damage prior to repair. A variation of the residual strength test is to load the structure to a predetermined load and to fire projectiles into the structure to determine if the built-in crack stoppers will arrest the damage incurred. Figure 6 shows the rapidity of failure incurred by a fighter wing with milled, integrally stiffened skin and no crack stoppers when hit by a single tumbling projectile 0.30 in. in diameter and 10 in. long.

Dynamic Tests

The aircraft industry initially relied solely on static testing and flight testing to prove the integrity of the airframe. However, as performance increased and structures became more efficient, it soon became apparent that the static simulation of dynamic conditions by the use of estimated inertia loads to establish equilibrium would not be possible for many design conditions. Hence some dynamic tests were devised to verify or predict the structural behavior or both.



CRACK PROPAGATION OF RIGHT WING

FIG. 6—Wing failure due to projectile penetration.

In vibration testing, the major structural assemblies—wings, tail surfaces, and fuselage—are excited by oscillatory loads to determine their resonant frequencies, mode shapes, and nodal lines. These data are used for the prediction of critical flutter speeds which must be sufficiently high so as to ensure that flutter is not experienced within the performance envelope.

The second category of dynamic tests usually performed is drop testing of landing gear struts to determine the energy absorption characteristics of the oleo struts and tires. These tests are performed to provide data for redesign of the orifice and metering pin combinations in oleo struts so as to maximize the energy absorption in the strut and reduce the forces applied to the airframe. Because carrier landing requirements of Naval aircraft are considerably more severe than for land-based aircraft, the Navy requires that the entire aircraft be drop-tested. In these tests, the desired gross weight and weight distribution for each landing configuration are simulated and the test aircraft is dropped from successively higher heights until structural failure occurs. Various landing attitudes are investigated and wheel spinup is used to simulate relative velocity between the wheel and ground and thus introduce drag loads into the structure. Since the entire aircraft is dropped, we obtain more realistic data on the energy absorbing characteristics of the tire, oleo strut, and aircraft structure, than is obtainable from the drop tests of the landing gear structure alone. Notwithstanding the complexity of these tests, they still do not adequately simulate actual carrier landings; to my knowledge no carrier based aircraft has successfully passed carrier qualification trials without further redesign of the landing gear or carry-through structure or both.

Recently, with the advent of closed loop testing systems, a test has been developed to simulate the dynamic effect of the rupture of the hold-back link during catapult launch of carrier based aircraft. For this test, the actual hold-back rupture link is installed; the appropriate engine thrust, catapult tow, and hold-back loads are applied; and provisions are made for application of appropriate inertia loads. The loads are increased gradually until the hold-back link suddenly ruptures and the entire aircraft moves forward; the closed loop system maintains the appropriate thrust and catapult tow loads and the inertia loads are permitted to build up to attain equilibrium and stop the forward motion with a limited forward travel.

The Operational Environment

Thus far I have discussed the use of essentially monotonic static tests and allied dynamic tests in establishing the structural integrity of the airframe for the highly improbable loadings to be encountered during the aircraft service life. These tests, other than the vibration tests, are char-

acterized by large scale plastic deformation and the resulting internal stress and strain redistributions above the yield strength of the structure. Hence, they allow poor detail design and poor quality control in manufacture. These tests are representative of the "safety-factor" concept of design where "never exceed" limits are established for operation of a machine, and a safety factor is applied to these restrictions to account for an occasional inadvertent overload and the material behavior in the inelastic range.

We are all aware of the fact that our aircraft normally operate at loads far below these "never exceed" limits and only very occasionally approach these limits. And yet we still get structural failures and near-failures that are discovered upon inspection. These are primarily structural fatigue failures.

Fatigue is a relatively new concept in structural design, although it was considered in mechanical design for many years. Prior to the end of World War II, all aircraft structures were designed using the "safety-factor" concept. No consideration was given to fatigue. Even though structural fatigue failures occurred in some of these aircraft, the failures were minor, progressed slowly, could easily be found on inspection, and usually occurred in the aircraft with considerably longer service than the average of the fleet. However, with the advent of higher strength materials having equal or less toughness than the ductile materials previously used, we had a spate of in-flight fatigue failures culminating in the much publicized Martin 202 and Comet disasters. Designers were still reluctant to refer to these failures as fatigue and used various other names such as "repeated load." Nevertheless, they were true fatigue failures and were caused by repeated applications of cyclic stresses far below the stresses resulting from the "never-exceed" loads used for design. This was vividly demonstrated during the many very simplified fatigue tests performed throughout the aircraft industry about 1950.

A new qualitative design requirement had been established, but how were we to implement it? How long should these structures last? For what loads? How could we prove the life in a test laboratory? How should we design structures to minimize the fatigue problem? Here was a new set of problems thrust on the aircraft industry busily engaged in making the transition from reciprocating engine aircraft to costly high performance jet aircraft.

The first question was easy to answer. For commercial aircraft the life could be established on an economic basis and has been set as high as 50 000 to 60 000 flight hours. The lives of military aircraft are established by economic as well as military requirements. Table 3 shows the Navy's design lives for various types of aircraft.

The second question—for what loads?—could not be readily answered for all types of aircraft. For commercial aircraft, which are designed

TABLE 3—*Service life for structural design.*

NAVY Service life for structural design					USAF Maneuver load spectra	
Navy types		Flying hours	Flights and landings	Flight maneuver spectrum	USAF types	Flight maneuver spectrum
VF	Fighter	3 000	3 000	A	Fighters F _I F _{II}	A
VA	Fighter-bomber	3 500	3 000	A	Fighter-Trainers	A
VT	Adv. Trainer	4 000	7 000	A	Bombers B _I	B
VA	Attack	4 000	3 500	B	Bombers B _{II} B _{III}	C
VS	ASW (CV)	4 000	9 000	C	Liaison	C
VS	ASW (land)	7 500	10 000	C	Trainers	C
VP	Patrol (sea)	5 000	5 000	B	Cassault	D
VP	Patrol (land)	7 500	10 000	B	Cargo	D
VC	Utility	7 500	10 000	C		
VT	Trainer	7 500	30 000	C		
	(other than Adv.)					
VW	AEW (CV) ^a	7 500	5 000	D		
VW	AEW (land)	20 000	15 000	D		
VR	Transport	20 000	15 000	D		

^a For carrier based (CV) airplanes, including carrier based trainers, the distribution of the total number of landings shall be as follows:

Carrier—20 percent FCLP—30 percent Field—50 percent

primarily for gust requirements since maneuvers are usually very gentle, we had the results of an NACA long term survey defining the frequency of occurrence of gust severity at various altitudes and over various types of terrain. These surveys are still continuing today. In addition the Air Force has conducted surveys to determine the frequency and severity of gusts at the extremely high and low altitudes used by military aircraft. But military aircraft are designed for maneuverability as well, and for fighter and ground support aircraft the maneuver loads are far more critical than the gust loads. Practically no reliable data on the frequency of occurrence of maneuver loads existed. Furthermore, equipment for obtaining the large amounts of data required for establishing statistically valid load-frequency curves was not available at the time. In spite of a joint Air Force-Navy-NASA program, reliable magnetic tape equipment for measuring and recording the desired data is still not available at a cost and in the quantity required.

In the interim both the Air Force and Navy set about to develop very simple systems to measure and count the frequency of occurrence of normal cg accelerations, which is the primary parameter in the determination of wing loads. The envelope of the early data was used to establish the initial load factor-frequency of occurrence data (Table 4) used for current design. Large numbers of counting accelerometers (Fig. 7) are installed in Navy aircraft and continue to supply data to be used for updating the requirements when necessary. Some typical data are shown in Figs. 8 and 9. Note that in the latter figure, the actual usage data show that the aircraft is used more severely than it was designed for and hence

TABLE 4—Frequency of maneuver loads.

Percent of maximum (positive) symmetrical limit load factor (column 3 of table I of Specification MIL-A-8861)	Flight maneuver load spectrum			
	A	B	C	D
35	17 000	25 000		
45	9 500	10 000	10 000	
55	6 500	3 500	3 000	1 000
65	4 500	1 000	1 000	150
75	2 500	500	300	20
85	1 500	200	100	3
95	300	75	30	0.5
105	150	25	10	0.05
115	40	10	3	
125	16	3	2	
Percent of minimum (negative) symmetrical limit load factor (column 4 of table I of Specification MIL-A-8861)				
0	500	5	0.7	
10	200	2	0.5	
20	100	1	0.25	
30	60	0.25		
40	35			
50	30			
60	25			
70	20			
80	15			
90	10			
100	5			
110	3			

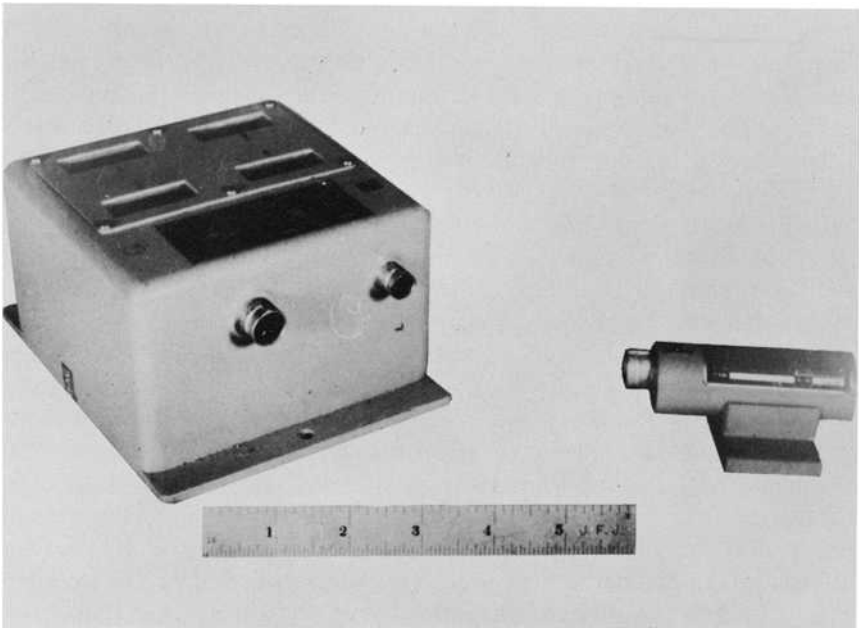


FIG. 7—U. S. Navy counting accelerometer.

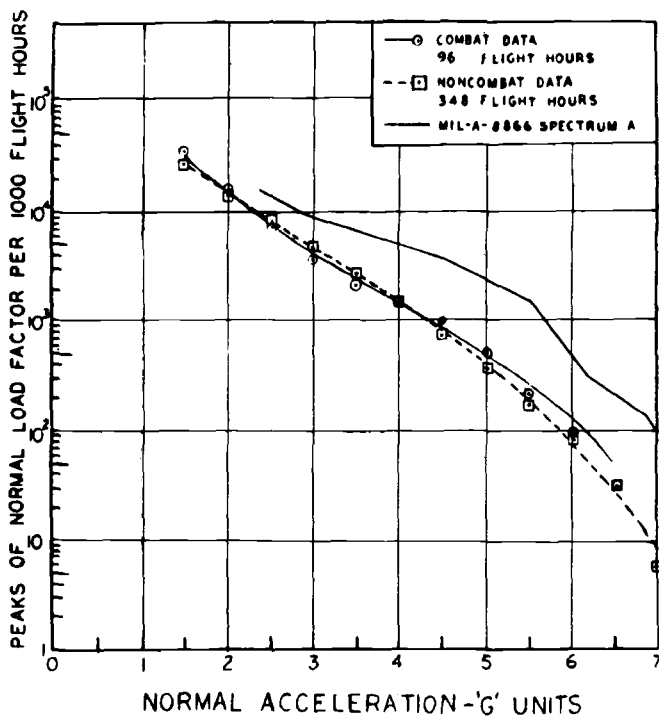


FIG. 8—Frequency of load factor exceedance curves—light attack airplane.

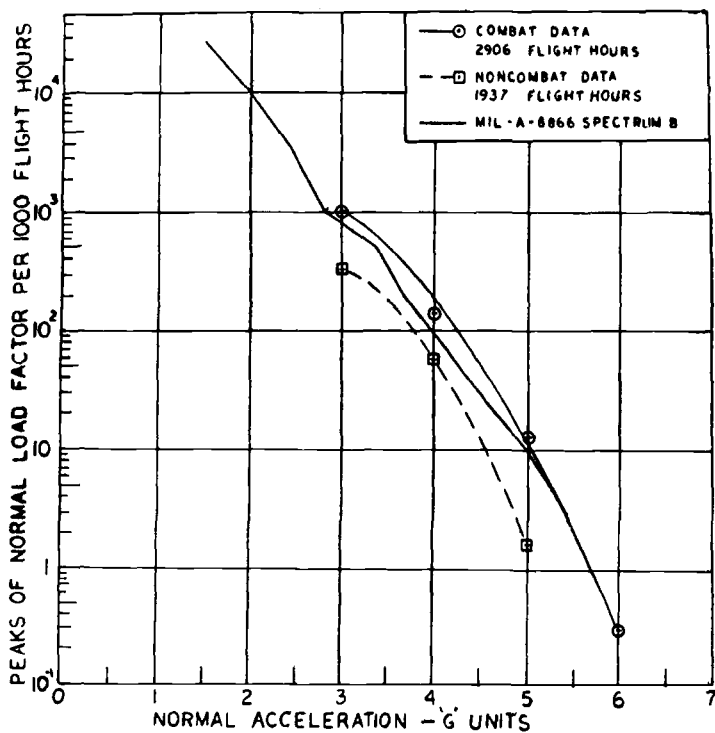


FIG. 9—Frequency of load factor exceedance curves—heavy reconnaissance airplane.

the life is shortened. Since this simple system measured only the cg acceleration, no data are available for determining the frequency of occurrence curves for tail loads; hence specific requirements are not included in the specifications. Other instrumentation equipment is currently being developed that can be used to rectify this deficiency.

For determination of the frequency of occurrence curves for landing loads, we use a photographic technique. From the film strips we can determine forward velocity, sinking speed, altitude, and sideslip velocity. Probability curves of sinking speed occurrence are contained in the specifications for carrier landings, field carrier landing practice, land plane landings, and training. The actual sinking speed frequency of occurrence curve for design of a particular airplane is based on the proportionate amount of time we expect to use the airplane for each type of landing.

The temperature environment has not been a problem to date because service aircraft do not fly at high speeds for sufficient lengths of time to raise the skin temperatures significantly. Only in the latest aircraft; namely, SST, Concorde, and new high performance fighters, will temperature be a significant factor other than for the structures around engines and other mechanical equipment. The temperature environments used for design of these new aircraft have been derived from data obtained from high performance research aircraft. The corrosion environment is known but cannot be quantitatively defined.

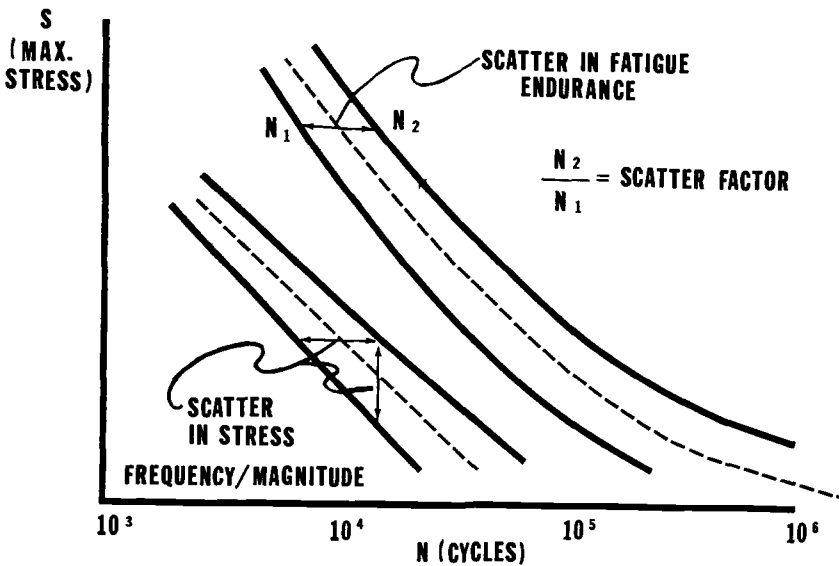
The Design Stage for Fatigue

It is the problem of the structural designer to satisfy the operational environment described in the detail specification for the airplane. In this phase of the work, selection of materials best suited and most fatigue resistant for the particular job, and the simplest, most direct load paths are essential in obtaining an efficient structure. During the cut-try-analyze period in developing the preliminary structural design, static load stress analyses predominate because of the labor entailed in preparing a formal fatigue analysis. Intuitive engineering and past experience are used to preclude many of the small errors which precipitate fatigue failures. If failure of any one part might generate a complete or catastrophic failure of the airframe, either some fail-safe feature, such as redundancy, crack "stoppers," or secondary structure to carry load after an initial failure, must be introduced or a safe-life design must be developed with a life expectancy well beyond planned usage.

Once the design is "frozen," complete formal stress analyses are undertaken for every major structural member, fitting, and connection. The analyses are performed for both static-dynamic load and fatigue conditions. The static-load analysis, based upon minimum guaranteed material mechanical properties, provides some assurance that the planned mission

can be performed with safety and can be readily verified during the static test program.

For the safe-life or fatigue analysis, there are no precise methods available to the designer even though we hopefully assign life limitations to components or entire airframes. There are available to us several hypotheses, and variations thereof, which presume that most materials in structures will behave in accordance with some mathematical rules which have been derived empirically. I do not intend, in this presentation, to review the current status of methodology used in analytical prediction of fatigue life, crack propagation, or residual strength. An excellent review paper and bibliography covering these topics was recently presented by Hardrath[1]. The objective of this entire program is depicted in Fig. 10. The dashed line on the right represents the average life of a material, a component, or a complete structure determined from constant nominal stress amplitude tests of a large number of specimens. Since there is usually considerable scatter in the data, the behavior is represented by a band bounded by the two solid lines which represent the S-N behavior with given probabilities of occurrence. Similarly the left hand band represents the variability of the stresses applied. To achieve a satisfactory structure, the two bands should not overlap. This diagram can also be used to depict the behavior of a material, component, or structure subjected to a spectrum of loads if we consider the ordinate to be the maximum stress applied in the spectrum and the abscissa can be total number



N=No. of cycles to failure or No. of times stress S is reached or exceeded

FIG. 10—Diagrammatic representation of endurance curve and spectrum.

of cycles, number of repetitions of the spectrum, or equivalent number of simulated flight hours.

From this discussion, one might conclude that by skillful use of sophisticated analytical techniques an eminently reliable airframe could be manufactured. This is far from the truth. It has been estimated that more than 90 percent of all fatigue failures in structures and machine parts are caused by faulty detail design or manufacturing defects. The faulty design involves such items as improper manufacturing procedures, inadvertent stress concentrations, and ignoring such items as fretting and corrosion, which can only be considered qualitatively.

Even deficiencies in maintainability have contributed to structural unreliability. The ability of the user or operator to maintain the airframe with ease, to inspect and make repairs, to restore damaged or deficient protective finishes, and the accessibility needed for such maintenance functions play an important part in reaching the design life.

While not considered as design parameters, the designer has an obligation to consider the effects of all potential causes of premature failures. They can render his predictions useless—they have been known to cause failure in less than one-tenth of the predicted life. The designer must participate in establishing the levels of quality required in manufacture of the airframe. From then on, it is the obligation of manufacturing management to police the inspection, training, motivation, supervision, and all other elements essential to maintaining the required levels of quality control. Otherwise the best design efforts go for naught, and stress and fatigue analyses become academic exercises.

Predictive Testing in the Design Stage

Predictive testing plays an extremely large part in the design stages. Figure 11 depicts what we would like to do—predict the airframe structural behavior from a knowledge of the material behavior—and shows the various steps necessary for this prediction. With the present state of knowledge, we can predict the behavior from one step in the chain to the next with an acceptable amount of error, but when we attempt to predict two steps ahead, the errors multiply and become totally unacceptable.

In metal structures we start at the first step in the chain which is a knowledge of the mechanical and fatigue properties of the material to be used, and then predict the static strength and fatigue life of each component, whether it is a fitting, a joint, a stiffened panel, etc. When considered desirable, component static and fatigue tests are performed and design changes made where required. In the next step, typical structural subassemblies, such as a section of the wing box beam or of the fuselage, are assembled, tested, and the design is further modified as required.

These component and subassembly tests enable us to verify and modify the design prior to fabrication of the complete airframe. How-

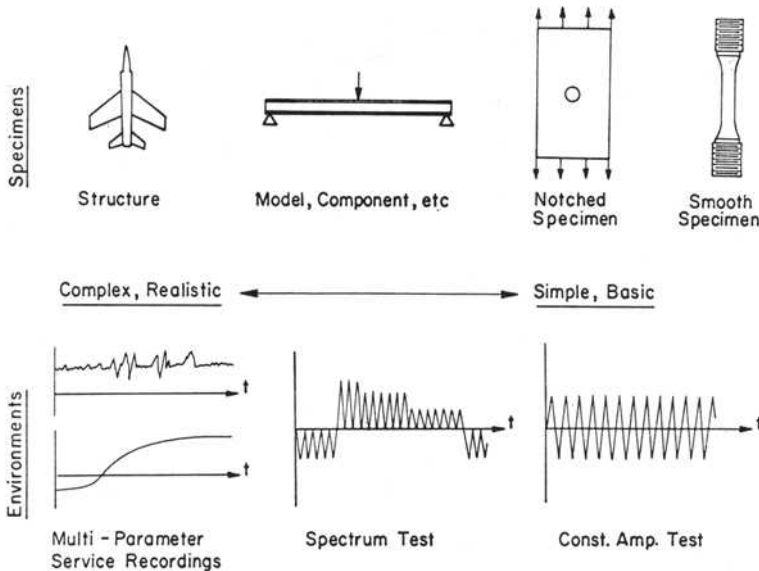


FIG. 11—Fatigue testing and service simulation.

ever, they do not necessarily enable us to predict the behavior of the complete structure because the test boundary conditions are established to try to duplicate the boundary conditions which were assumed in the stress analysis to make it mathematically tractable. Whether these assumptions are conservative or unconservative cannot be established with certainty until the complete airframe is tested.

Testing the Complete Airframe

The load and structural analyses performed are subject to the mistakes and inadequate assumptions associated with human limitations. Fortunately, many of these deficiencies can be corrected or controlled by a combination of laboratory and flight investigations during which functional, structural, and configurational tests can be made. The complete airframe is static and dynamic tested for load distributions representative of a variety of take-off, flight maneuvering, and landing conditions to investigate the validity of structural analyses and to prove that the airframe is indeed ready for flight. These tests were discussed earlier.

These tests usually disclose a host of changes needed for the production version of the test prototype. Because of the unpredictable effects of many of these changes on structural life, fatigue tests are usually deferred until a representative production airframe is available. Only those parts of the airframe considered critical for mission safety and planned life are fatigue tested; for example, only the wings and parts of the fuselage and tail of an airplane may be fatigue tested. Because of the serious economic

results of landing accidents and loss of military potential, repeated drop tests are now required by the Navy to disclose fatigue life and deformation inadequacies.

The fatigue tests are performed, usually on a virgin airframe, only for those conditions deemed critical. In the case of large components, only one fatigue test is performed, and a scatter factor to take care of the variability in structural response is applied to the final results. In the Navy this scatter factor is usually 2 since we use a very severe load spectrum. The Air Force uses a less severe spectrum and requires a factor of 4. At the first evidence of a failure, an attempt is made to rework the area or reinforce it in some fashion to reduce the stress concentration. If possible, element tests are performed separately to prove out the change without risking catastrophic failure of the entire structure.

The methods used for applying load to the airframe are the same for fatigue tests as for the static tests discussed earlier. Figure 12 shows the setup for the fatigue test of the wing of an advanced trainer. The main difference between the setups for the two types of tests is that the static test need only be set up for one loading condition at a time; for a fatigue test a number of load setups for different loading conditions must be set up simultaneously and provision must be made for automatic switching from one loading condition to another. Consider, for example, the test of

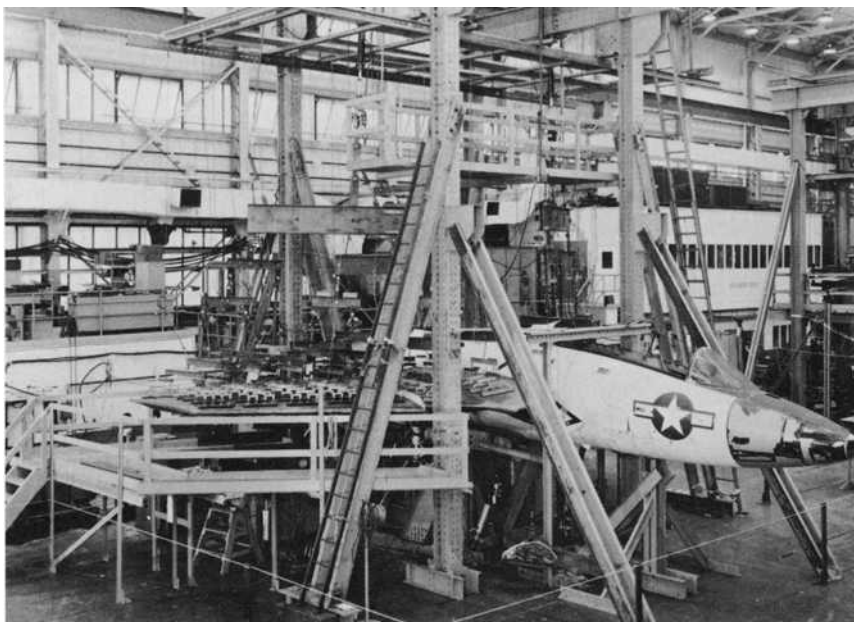


FIG. 12—Wing fatigue test setup—advanced trainer.

a commercial transport; the following loads, including cabin pressurization as required, must be applied to simulate a single flight:

1. Taxi: Landing gear loads plus inertia loads due to runway roughness.
2. Take-Off: Wing, tail and fuselage net loads and loads on auxiliary lift devices such as flaps and slats.
3. Climb: Wing, tail and fuselage net loads due to maneuver and gust.
4. Cruise: Wing, tail and fuselage net loads due to maneuver and gust.
5. Descent: Wing, tail and fuselage loads due to maneuver and gust.
6. Approach: Wing, tail and fuselage net loads and loads on auxiliary lift devices, such as flaps and slats.
7. Landing: Landing gear loads plus wing and fuselage net loads.
8. Taxi: Same as (1).

For a Navy carrier based, swing-wing fighter we would get a similar breakdown with the additional complexity of including catapult and arrestment loads in the take-off and landing phases and appropriate wing sweepback during the climb, cruise, and descent phases. Fortunately we can simplify the testing of such an aircraft. First, taxi loads are negligible in comparison with the other loads; hence, they can be omitted. Second, catapult and arresting loads are usually critical only locally; hence, they can be applied during separate tests. Third, the take-off and approach condition loads would not be critical except for flaps and slats, which can be tested separately. This leaves only the climb, cruise, and descent conditions, which can be combined into a single loading condition for each sweepback position, and the landing condition. Reorganizing in this manner permits simultaneous performance of two or more tests and thus expedite the program.

What Do We Learn from the Airframe Fatigue Test?

Since analytical prediction of fatigue life leads to highly questionable results, we conclude that the best estimate of expected life is obtained from the fatigue test of the complete airframe for programmed loads simulating those expected in service. But only if the test is well designed and consideration is given to the many significant factors, such as load sequence, spectrum truncation, negative loadings, etc. Many times economic considerations cause us to modify the load sequence, omit negative loads, and truncate the low load end of the spectrum. These compromises introduce systematic errors in the expected life, usually in the unconservative direction. However, the compromised test will still locate the weak details in the structure and assist in evaluating the modifications incorporated. This may not be as undesirable as it may seem at first

glance, at least for military aircraft. These aircraft, with their variable missions, seldom fly in accordance with the design load spectrum. Figure 13 shows the variations in usage determined for four different fighter aircraft. The differences result from mission changes after delivery of the aircraft. Hence the life estimate as determined by the uncompromised test program may be in just as much error as the estimate obtained from a simplified test. This observation is not valid for civil aircraft, which are designed primarily by gust and landing requirements. However, even here there is a significant difference in loads experienced between two identical aircraft flying consistently over different routes. What, then, is the value of full-scale fatigue testing?

The answer to this depends upon the intended use of the aircraft and the design philosophy employed. Military aircraft are designed primarily for safe-life; that is, no failure should occur during the planned useful life of the aircraft. On the other hand civil aircraft, which are designed for a much longer life than military aircraft, are customarily designed utilizing the fail-safe philosophy. In actuality neither philosophy is used exclusively. Fail-safe features are included in military designs whenever operational use and performance would not be degraded, and safe-life features are included in civil aircraft whenever redundancy in structure cannot be incorporated. In the latter case the safe-life is usually a large enough multiple of the design life to preclude failure.

In the case of safe-life structures, the fatigue test is essential from a safety standpoint so as to confirm the airworthiness of the structure. Minor failures are repaired and major failures result in component re-

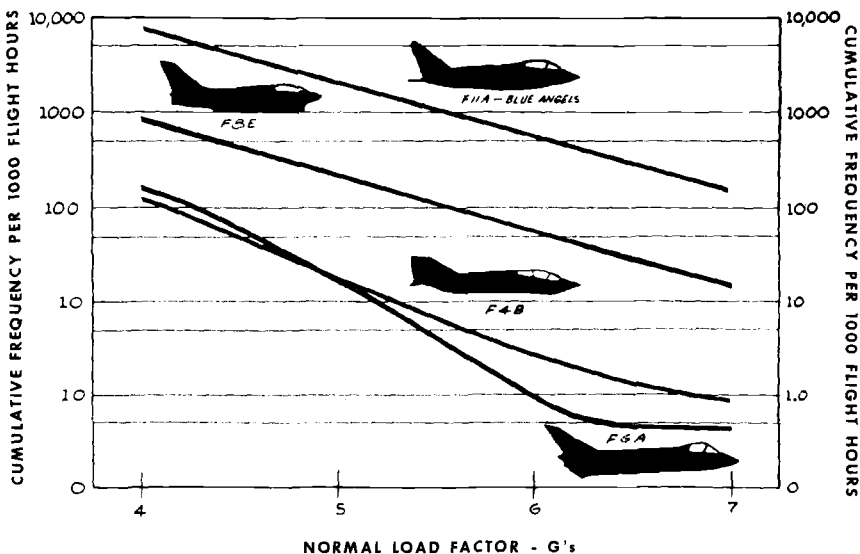


FIG. 13—Typical normal load factor experience of naval fighter aircraft.

design. The test is continued to catastrophic failure of each major structural assembly and the test life must exceed some multiple of the design life (multiple of 2 for the Navy and multiple of 4 for the Air Force for the specified load spectrum). Poorly designed parts and blatant manufacturing flaws usually show up long before the design life and changes can then be incorporated in future production and by service retrofit. However, manufacturing flaws are not necessarily typical in all "identical" parts; hence, it is possible that some flawed parts may be on production aircraft and not on the test article.

The uncertainties in anticipating service loadings, the inability to identify unrepresentative, but potentially dangerous, details by tests, and the inadequacy of analytical procedures indicate that the best index of the life of a structure presumed to have a "safe-life" is actual behavior in service. Fortunately most of these structures develop cracks in advance of failure which can be detected by periodic and careful inspections. Since the detectable crack propagation stage for a safe-life structure is usually very short, these inspections must be performed very carefully and with very short intervals between successive inspections. The areas to be inspected and the nondestructive test procedures to be used have usually been established during the tests.

Since it is virtually impossible to predict a safe life with satisfactory confidence even though a complete airplane is tested the manufacturers of civil aircraft have placed more emphasis on fail-safe construction and on periodic inspections and maintenance to assure a safe aircraft. The fail-safe philosophy depends on consideration of the residual strength of partially failed structures and of rates of crack propagation. These are features of the fatigue process that are all but ignored in safe-life structures. The fail-safe structure must be inspected frequently enough so that a crack that is not detected in one inspection will not grow to a length that will cause failure under normal operating conditions within the inspection interval. Thus the fatigue test of a fail-safe structure must provide data on when and where to inspect and information on the crack growth rates. In addition parts replacements or repairs to the structure must be evaluated. Because of the time and cost required to perform these tests for a large civil transport, termination of the test is usually by mutual agreement among the manufacturer, the purchaser, and the certifying authority.

Unfortunately, the variable results of structural fatigue testing can raise serious doubts as to the validity of life predictions. Because of the considerable variability in service loadings and the variability in fatigue test results, the results of a complete airplane fatigue test must be interpreted in light of the statistical nature of the basic inputs and responses. Even the reliability of the test method and the integrity of the tester are, at times, open to question. It is evident that prediction of endurance life is

still in its infancy and that fatigue tests can be performed to reveal almost any answer desired, even when the quality of the product tested is uniformly representative of production quality.

To demonstrate the validity of this statement, here is one case from the literature where merely altering the sequence of application of the loads had a significant effect on the test results. In a study of the effect of a number of parameters on the fatigue life of an aluminum alloy box beam in simple bending, Breyan[2] studied the systematic effect of changing the test block size on the beam life. Figure 14 demonstrates what we mean by block size; the loads within each block can be applied in either a fixed sequence as shown or can be randomized within each block. The test results are depicted in Fig. 15 for three different load spectra specified in MIL-A-8866. Note that changing nothing but the test block size and keeping the same load frequency distribution, the fatigue life of the structure can be affected by a factor of 2. Similar results were obtained when the load sequence within each block was randomized but the factor was reduced somewhat and the results were not as consistent. Thus by the simple expedient of specifying large block sizes, which makes the test less costly if more than one loading condition is used, the testing agency can demonstrate the structure for the desired life yet the demonstrated life may be significantly unconservative.

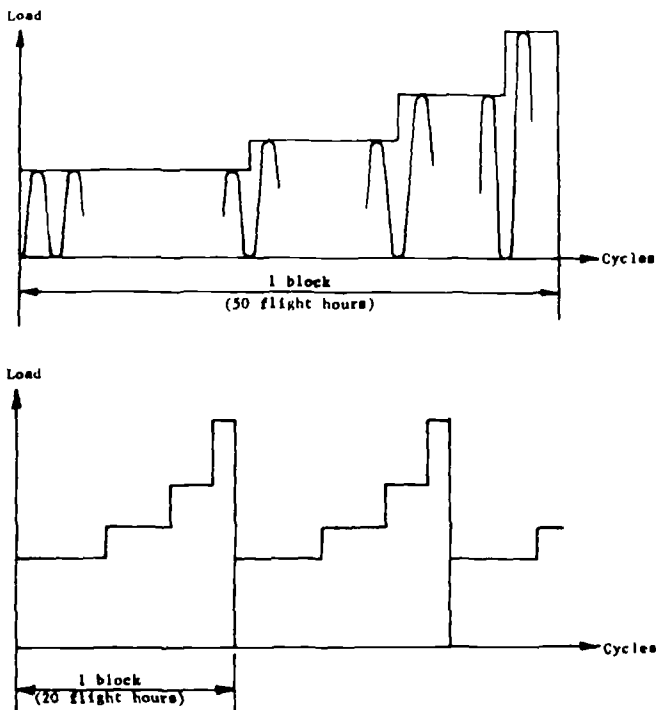


FIG. 14—Block size definition.

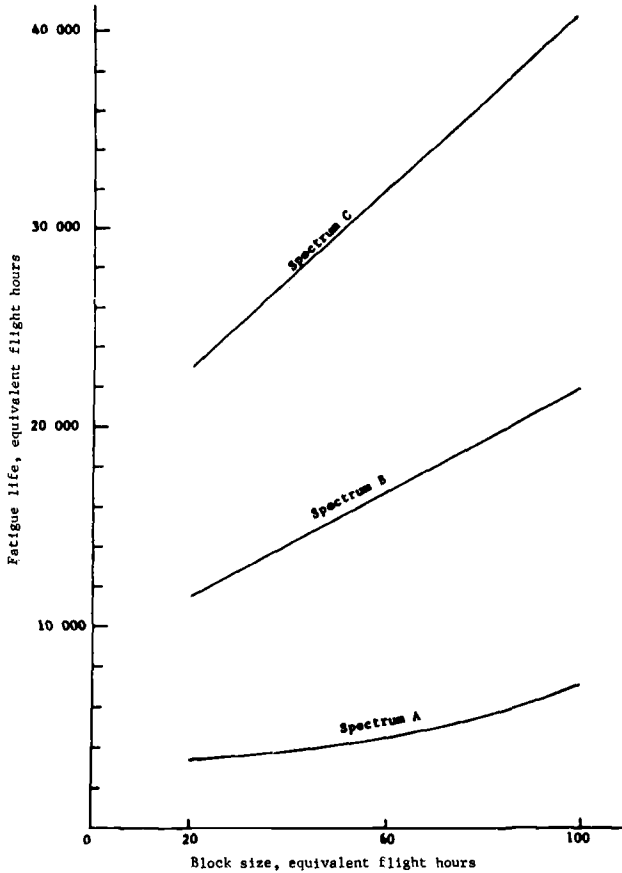


FIG. 15—Effect of spectrum block size of fatigue life.

Comparison of Test and Service Experience

Although everyone performing fatigue tests of aircraft structures tries to correlate test and service experience for aircraft he is familiar with, the correlations are usually not publicized. Troughton[3] compiled test and service data supplied by manufacturers for over 40 aircraft, both military and civil. Most of the information compiled was provided by British manufacturers. A similar compilation of the extensive information available for U. S. aircraft should be made.

Figure 16 summarizes the experience of nineteen aircraft types. The five aircraft on the right are important civil aircraft of recent vintage operating in large numbers. It is clear from this that modern design and testing techniques are giving much better results than in the past bearing in mind that these aircraft have considerably higher design target lives than the others. Other data in this paper show that the service defects

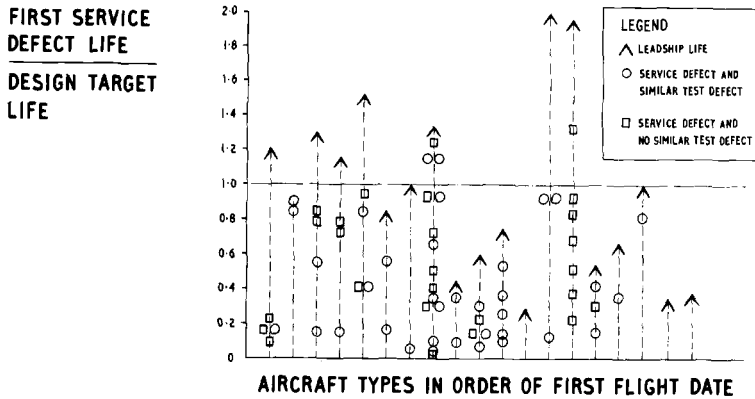


FIG. 16—Significance of aircraft service defects.

generally occur earlier than the corresponding test defects. This is related to the statistical nature of the problem since the test article probably represents the median experience and the first failures represent the lower threshold of the probability distribution for each failure.

Troughton also made an analysis of the reasons for the discrepancies between service and test (Table 5). Note that all the reasons, except for environment, have already been discussed.

Effects of Environment

Up to this point I have ignored any discussion of environmental effects on fatigue behavior. The two environments we are most concerned about are elevated temperature and corrosion. Both are time dependent effects and as yet we do not know how to superpose them on the cycle dependent fatigue behavior other than to run the entire test in real time. This is not economically feasible. If we assume that 50 percent of the elapsed test time is actually used for applying cyclic loads to the structure and the remaining 50 percent is used for inspection and repair of the test structure, then the maximum equivalent flight time we can apply to the test structure is less than 4500 h per year. For a civil SST transport with a desired service life of at least 50 000 flight hours, this would mean more than 11 years of real time testing. Furthermore, since the service aircraft would be accumulating flight time at approximately the same rate, the test should start at least 2 to 3 years prior to the first commercial flight to be on the safe side. Consequently, some method of time compression is required.

The feasibility of compressing the time of the simulated mission profile during elevated temperature fatigue testing has been under active investigation by both NASA[4] and the Air Force[5,6]. The results to date indicate that time compression during elevated temperature fatigue testing is feasible; however, the compression factor depends on the thermal characteristics of the structure. Considerable additional work is required.

TABLE 5—Reasons for discrepancy between service and test.

Reason	First Service Defect						
	No Test Defect	0.01 to 0.20	0.21 to 0.40	0.41 to 0.60	0.61 to 0.80	0.81 to 1.00	1.01 and Above
							1965 Total
1	Actual Flight Usage Different From Test Assumptions	6	2	3	...	...	...
2	Incorrect Loads Applied On Test	5	2	...	...	...	11
3	Test Loadings Not Representative	21	10	2	...	...	7
4	Test Specimen Not Representative	7	2	5	1	...	35
5	Environment Different: Local Heating, Corrosion etc.	5	1	1	...	...	15
6	Problems Highlighted by Service Inspections	1	1	2	...	...	7
7	Unknown	2	4	2	5	...	4
	TOTAL						95
	Total						52

The problem of the corrosive environment is at least in order of magnitude more difficult than the temperature problem. It has been estimated that the military and NASA together spend \$6 billion annually on prevention and control of corrosion damage. In Southeast Asia five manhours, or 20 percent of total maintenance time, are spent to prevent and repair corrosion damage for every hour of flight. In addition to this huge cost and maintenance manhour expenditure, we have the additional cost due to loss of use, which, in the military, must be compensated for by larger fleet sizes to maintain a given level of combat-ready aircraft.

The Navy has attempted to reduce the effects of corrosion on aircraft strength and life by developing corrosion resistant alloys, by developing protective treatments applied during manufacture and overhaul of aircraft, and by establishing some arbitrary design restrictions to inhibit stress corrosion cracking. However, the efficacy of these methods and procedures have been evaluated in an arbitrary manner that bears no relation to actual service use.

I recently participated in two laboratory test programs to evaluate the effects of corrosion on fatigue life. One, which has not yet been published, was an investigation of the effects of corrosion on the lives of simple, built-up, aluminum alloy box beams; the corrosion was introduced artificially by periodically spraying the bare specimens with a 3½ percent NaCl solution with H₂SO₄ added to get a $P_H = 1.5$ to simulate the effect of carrier exhaust stack gases. The effect on life was arbitrarily evaluated by running load spectrum tests on two sets of specimens—one set exposed to the normal laboratory environment and the other set periodically exposed to the corrodant. Corrosion reduced the average life of comparable specimens by as large a factor as 10.

The second investigation[7] was to determine the effects of surface pitting and exfoliation on the fatigue life of 7075-T6 aluminum alloy spar cap extrusions from an amphibious air-sea rescue aircraft. In this case the corrosion occurred naturally during the service life of the aircraft; when severe intergranular corrosion was indicated by ultrasonic inspection, the spar caps from two airplanes were replaced and the corroded spar caps were used for laboratory investigation. Large dogbone specimens were fabricated from the skin attachment flange of the spar cap containing a number of rivet holes; specimens were taken from areas with no corrosion, surface pitting only, slight exfoliation, and surface pitting with severe exfoliation. Both constant and variable amplitude tests were run with similar results; the constant amplitude test results are shown in Table 6. Using these data, the life remaining in a typical aircraft of this model was estimated. Since the estimated life remaining was insufficient, a fatigue test of a typical wing was conducted and the results confirmed the estimate. A fix was developed and tested prior to installation in the service aircraft.

TABLE 6—Constant amplitude fatigue test results for specimens from a corroded spar cap.

Specimen Condition	Specimen No.	Cycles to Failure
No Corrosion	8	1277
	9	1432
		1354 Avg.
Surface Corrosion Only	1	1500
	4	1308
		1404 Avg.
Slight Exfoliation	7	1252
Surface Corrosion And Severe Exfoliation	6	610
	10	896
	11	819
		775 Avg.
Mean Stress = 0		
Alternating Stress = 35 000 psi		
All specimens made from skin attachment leg		

The tests of the corroded spar cap specimens and the full-scale wing seem to provide sufficient data for life prediction including the effects of corrosion. These predictions are necessarily unconservative because the corrosion process was appreciably slowed down and possibly stopped when the test articles were removed from the corrosive environment. In actual service, the corrosion process would continue but there is no way of evaluating this additional effect.

Closing the Loop

A truly quantitative strength or life prediction for an airplane structure would require the assumption of a number of hypotheses, such as: (1) if the mission assumed for design does not change; (2) if the airframe configuration will not change after tests; and (3) if the operator obeys operating restrictions, and many more.

The quantitative prediction could, for practical purposes, be variable throughout the life of an airplane and be capable of precise prediction only after it had completed its service life and the facts concerning failures and usage were known with some degree of accuracy.

Information concerning failures and the life to failure are available from the tests performed during the design and evaluation phases although only for the configuration tested and for the loads used in the test. To interpret these results for individual aircraft, we must get data on the actual service usage of each aircraft. Initially service usage was defined only in terms of flight hours. This is not a good criterion, except possibly for civil transports, since aircraft may differ widely in the severity of individual usage. Obviously, then, we need some means for determining the severity of usage experienced by individual aircraft and for com-

paring this with the test data. One method for accomplishing this is to install a counting accelerometer in each aircraft as it is delivered and to obtain an index of usage severity from the data. These counting accelerometers are the same instruments previously discussed that are used for obtaining design load factor frequency of occurrence distributions. Thus the data serves a dual function, to obtain information for future designs and to monitor usage for damage estimation for individual aircraft so equipped. It is the Navy's policy to install these instruments in all new aircraft. The data from each instrument is recorded and analyzed periodically to obtain a damage index for each airplane which is then used for comparison with the test results.

It has long been realized that measurement of cg acceleration is but a crude approximation of wing damage since it is just one of the parameters determining the wing loads. Furthermore it provides no information for assessing the damage to other parts of the airframe. Within the past few years, considerable laboratory work has been done to develop and evaluate simple and inexpensive fatigue damage monitors. One such device is a strain counter; it is similar to the counting accelerometer except that a strain transducer instead of an accelerometer is used. The transducer is installed at a critical location on the structure and the number of exceedances of predetermined strain levels are recorded. An even simpler device, the fatigue life gage, is being critically evaluated. This gage is similar to an electric resistance strain gage except that it has been fabricated to produce a large zero shift (measured by resistance change) as a function of the strain excursions and numbers of cycles applied. Both the strain counter and the fatigue life gage would be installed on the fatigue test airframe as well as on each service airplane; the data obtained during the fatigue test would then serve as the basis of comparison for the service data.

Extension to Other Fields

This entire paper has been devoted to the role played by structural testing in the development of reliable airframe structures. It was intended to serve as an example of the extent to which one industry attempts to build safety and reliability into its structures. The methods and equipment used are equally applicable to other fields of structural design. The automotive industry wholeheartedly has adopted closed loop simulated service testing for complete vehicles and components and uses these testing methods for prediction of vehicle performance.

Other structural fields are slowly developing their own criteria and predictive test methods. The strain histories of large buildings and bridges are now being determined by the use of simple mechanical recording strain gages[8]. In the shipbuilding industry, there is an increasing awareness of the fatigue and reliability problem because of the introduction of high strength materials and the trend to lighter structures. This is es-

pecially evident in the development of deep diving submarine hulls and of high performance light ship structures required for hydrofoil and captured air bubble craft. Both the building and ship structural fields rely heavily on the use of scale structures for test and considerable work remains to be done to correlate the results of full-scale and scaled fatigue tests.

Concluding Remarks

In this paper we discussed the role played by structural testing in the development of reliable airframe structures. The three-fold safety-factor, safe-life, and fail-safe design approach provides reliable airframe structures when complemented by an effective laboratory test program that includes static, dynamic, and fatigue tests.

It should be evident from the remarks herein that operational strength limits can be predicted fairly accurately from the static test results. It should also be evident that prediction of the safe service life from fatigue test data alone leaves much to be desired. Nevertheless, the alarming cost of modifying aircraft in service has gained an almost complete acceptance of laboratory structural fatigue testing as an economical way of doing business.

The ability to predict the lives (actually, the lives remaining) of individual aircraft depends upon the development of an accurate fatigue damage monitor. Until that time, frequent and rigidly controlled periodic inspections must be carried out on all aircraft and an elaborate scheme of progressive preventive maintenance is mandatory.

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