**CRACK PROPAGATION CHARACTERISTICS
OF Zr-2.5 wt% Nb ALLOY TUBING**

P. J. PANKASKIE

OCTOBER, 1967

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OF Zr-2.5 wt% Nb ALLOY TUBING*

By

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October 1967.

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ABSTRACT

As part of the evaluation of Zr-2.5 wt% Nb alloy as a candidate pressure tube material, the crack propagation behavior of tubing was investigated.

The crack propagation behavior of Zr-2.5 wt% Nb alloy tubing, as determined from tests using milled slots as stress concentrators, is not significantly altered by cold working or heat treatments from room temperature to 300 °C. When hydrided to 250 to 300 ppm of hydrogen, the fracture strength of a defected tube is generally reduced and its crack propagating behavior tends to be less predictable.

Fatigue testing of tubes in the tension-tension mode by cyclic internal pressurization at peak cyclic hoop stresses of about one-fifth of ultimate hoop strength was continued until fatigue crack growth and catastrophic failure occurred. It appears that under these loading conditions, leakage through a fatigue crack is likely to occur when the fatigue crack dimensions are appreciably smaller than the critical size for catastrophic failure. Data obtained to date suggests that the critical crack length for catastrophic failure may diminish with increasing amounts of prior fatigue damage.

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CRACK PROPAGATION CHARACTERISTICS
OF Zr-2.5 wt% Nb ALLOY TUBING

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INTRODUCTION

Within the United States and Canada, there are several water-cooled thermal reactors that use pressure tubes in the reactor core section of the primary coolant system. Additional such reactors are in construction or planning stages. Because of the small thermal neutron capture properties of aluminum and zirconium, their alloys are prime candidates for pressure tubing in reactor cores. For power producing reactors, only the zirconium alloys possess sufficient strength at intermediate temperatures (e.g., to 400 °C) to be of engineering and economic interest.

One of the objectives of the USAEC-AECL Cooperative Zr-Nb Alloy Program is to evaluate the crack propagation behavior of Zr-2.5 wt% Nb alloy for comparison with Zircaloy-2.⁽¹⁾ This report presents the data obtained to date and briefly outlines areas for further research and evaluation.

Presently, the Zr-2.5 wt% Nb alloy appears amenable to some combination of cold working and heat treatment that may improve strength properties, and for this reason, this alloy is being evaluated as a candidate material for the fabrication of pressure tubes for several water-cooled reactors now in the planning stages.

Creep deformation, corrosion, and crack propagating behavior of candidate materials are important considerations in the design of pressure tubes for reactor core cooling. Creep and corrosion of the tube material are commonly recognized as the factors limiting service life. Since creep and corrosion are

inherently slow processes, most reactor pressure tube design criteria allow for both. Even under adverse service conditions, which may cause accelerated creep and/or corrosion, there usually is some time available in which to correct operating conditions or replace the tube without excessive risk to the reactor core structure. However, the situation is different when the brittle fracture problem is considered. Brittle fracture can result in total destruction of a pressure tube and cause considerable damage to surrounding structural components; the occurrence of it can have rather far reaching consequences. Hence, the conditions under which a flaw may be at critical size or grow to critical size resulting in unstable fracture is important in the design of pressure tubes.

SUMMARY AND CONCLUSIONS

As a part of the program to evaluate Zr-2.5 wt% Nb alloy as a candidate pressure tube material, crack propagation tests, employing both milled slot and fatigue crack defects, were done.

Tests using a milled slot as a stress riser and crack starter showed that there is no significant difference in the crack propagation behavior of 30 and 60% cold-worked and heat-treated tubing when the crack propagation results from internal pressurization at temperatures ranging from room temperature to about 300 °C. When hydrided to concentration levels of 250 to 300 ppm hydrogen, the fracture strength of a defected tube is generally reduced and the crack propagation behavior tends to be unpredictable. Comparisons of defected Zr-2.5 wt% Nb alloy and Zircaloy-2 tubing show that there is little difference in fracture strength and no significant differences in crack propagation behavior.

As compared to published data for sheet material, the stress intensification factors calculated from milled slot test data appear to be quite large, indicating the K_{IC} data cannot

be obtained from this type of test. Nevertheless, these tests show that a "blunt type" crack or defect can result in brittle crack propagation in tubular geometries under appropriate stress loading.

Fatigue testing by cyclic internal pressurization of tubular specimens can, in the presence of an appropriate stress riser, produce fatigue crack growth and catastrophic failure. The fracture toughness of a tubular specimen with a crack grown under cyclic loading conditions is substantially less than the toughness of a tubular specimen with a milled slot tested under static load conditions. Furthermore, the test data obtained suggest that at a given stress, the critical crack length for catastrophic failure decreases with increasing amounts of fatigue damage accumulated prior to failure.

These tests have shown that under fatigue loading, a flaw is likely to grow to a size at which leakage will occur prior to catastrophic failure. For peak cyclic stresses, which are about one-quarter or less of the ultimate strength, the leakage crack length is appreciably less than the critical crack length for catastrophic failure.

DISCUSSION

TEST SPECIMENS AND PROCEDURES

To obtain crack propagation data, two types of tests were employed. One type consisted of bursting by internal pressurization a tubular specimen with a slot milled into the outer surface. The depth of this slot was equal to about 80% of the tube wall thickness and the width was one-sixteenth of an inch. This type of test had been used for most of the testing of Zircaloy-2 properties and was used with the Zr-2.5 wt% Nb alloy so as to establish a valid basis for comparison of fracture characteristics.

The second series of tests, not yet complete, involves fatigue testing by cyclic internal pressurization of slotted tubular test specimens. This series of tests was initiated to obtain crack growth information; determine whether a pressure tube is likely to leak detectably before catastrophic failure; and provide more accurate data for calculating stress intensification factors.

Test Materials

Tube specimens were tested in 30% cold-worked, 60% cold-worked, heat-treated, and heat-treated and aged conditions. The cold-worked tubing (2.01 in. ID by a nominal 0.180 in. wall thickness) was produced from magnesium reduced zirconium sponge by extrusion and tube reducing. The heat-treated tubing (2.01 in. ID by a nominal 0.175 in. wall thickness) was produced from sodium reduced sponge by extrusion and cold drawing. The heat treatment consisted of water quenching from 880 °C and then cold drawn to a reduction of 15%. Some test specimens also were given an aging treatment which consisted of holding at 500 °C in vacuum for 24 hr.

Milled Slot Tests

For Zircaloy-2 pressure tubing, much of the crack propagation data were obtained from tubular test specimens in which a 1/16 in. wide slot was milled in the outer surface to a depth of about 80% of the tube wall thickness.⁽¹⁾ To make the data comparable, milled slots of the same width and depth were used for the Zr-2.5 wt% Nb alloy tubular test specimens. Tube crack propagation tests were done at temperatures ranging from room to 300 °C with the material in the following conditions:

- 30% cold-worked temper
- 60% cold-worked temper
- Heat-treated by 880 °C soak, water quench, and 15% cold-worked

- Heat treated, same as 3, plus aging at 500 °C for 24 hr
- Each of the above hydrided to an average hydrogen concentration of about 250 to 300 ppm.

For the crack propagation tests, the tubular test specimens were sealed and then pressurized until the remaining wall thickness at the milled slot failed. Except for a very few instances, failure was limited to the remaining slot web; there was no crack propagation. Each of the tubular test pieces was then fitted with an elastomeric plug to avoid excessive leakage without altering the stress loading significantly, and the test pieces were repressurized until failure. The general tube assembly ready for pressurizing is shown in Figure 1.

The elastomeric plug technique works satisfactorily at temperatures below that at which the elastomer softens excessively. For most elastomers tried, the upper useful temperature limit is about 150 °C, or slightly higher. Also, there was apparent degradation of the elastomer at temperatures in the neighborhood of 300 °C suspected, at least in part, to result from the high pressures involved. To circumvent this problem of elastomer degradation at the higher temperatures, a thin wall "liner" tube was employed to seal the web crack in some of the elevated temperature tests. Zircaloy-2 tubing proved to be the most satisfactory, and, taking into account the added wall thickness, provided data comparable to that obtained with elastomers. The maximum tube wall thickness of the "liner" tubes used was 0.030 in.

Fatigue Crack Tests

Fatigue tests were initiated to determine what differences existed in crack propagation behavior of milled slot crack starters and fatigue cracks. Furthermore, fatigue crack growth could show whether or not a crack large enough to leak would propagate catastrophically.

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FIGURE 1. Tube Test Assembly Ready for Pressurizing. (In the foreground is an exploded view of end caps.)

In these tests, a simulated surface flaw was made at the outer tube surface to produce a point of stress concentration during cyclic internal pressurization and to initiate a fatigue crack. Since no fatigue data for Zr-2.5 wt% Nb alloy tubing was found in the literature, and since it was thought that the endurance strength would be at least 20% of the ultimate static strength, a nominal peak hoop stress of 23,000 psi, or about 20% of the ultimate hoop strength, was selected. This corresponded to a peak cyclic pressure of 4000 psig for the tubular test specimens available. The minimum pressure used was about 100 psig. In the tests completed so far, the pressure cycling rate was about 400 cycles per hour. Cyclic pressurization was achieved through the use of an air operated diaphragm pump. With this type of pump, roughly 30 pump strokes are required to reach the peak pressure of 4000 psig. Under this mode of operation, pressurizing accounts for 90% of the time required to complete each cycle.

Slots with a semicircular profile produced by electrical discharge machining (EDM) have been used for the tests completed to date. The radii of the EDM slots were chosen so that the depths of these slots were about 70%, 55%, 40% and 30% of the wall thickness of the tubular test specimen. The fatigue tests completed to date were done only on heat treated material in both the aged and nonaged condition. Tests have not been applied to uniformly hydrided tubular test specimens.

DISCUSSION OF TEST RESULTS

Milled Slot Tests

The fracture strength of the tubular test specimens as a function of milled slot length are shown in Figures 2, 3 and 4. So the data obtained for all test temperatures could be plotted on a single curve, the ratio of hoop strengths of a slotted to

an unslotted test specimen was used rather than the actual fracture strength values for each test. The use of this ratio has little effect on the data scatter for all of these test data. A few data points were omitted since fracture strengths for comparable unslotted test specimens were not obtained for some material conditions at certain test temperatures. All test results are tabulated in Tables I and II.

Examination of the graphs in Figures 2, 3, and 4 show that the fracture strengths of the 30% CW, 60% CW and heat-treated materials are comparable. The nearly equal slopes of these curves suggest a greater similarity in behavior than was actually observed. When tested at room temperature and lower, the 30% CW and 60% CW test specimens rarely exhibited crack propagation immediately upon failure of the web remaining at the milled slot. For the heat treated test specimens, at the same test temperatures, crack propagation often occurred immediately after failure of the remaining web at the milled slot. This latter behavior is also characteristic of Zircaloy-2 at room and lower temperatures.

Considering the cold-worked and heat-treated Zr-2.5 wt% Nb alloy test materials, the heat treated condition generally has a slightly greater unslotted ultimate strength. Since the strength of slotted test specimens is about the same, this behavior difference must be due to metallurgical differences. Metallographic examination of the 30% CW, 60% CW, and heat treated test specimens showed minor differences in hydrogen content. As estimated by metallographic methods, the hydrogen content of the 30% and 60% cold-worked material was about 10 ppm, and for the heat treated it was about 30 to 40 ppm. Drop weight tests showed the nil ductility temperature for the 30% CW to be about -10 °F; for the 60% CW, to be -60 °F or less; and for heat-treated material, about -30 °F. Based on these results, the difference in behavior of the cold-worked

TABLE I. Crack Propagation Tests of Cold Worked Zr-2.5 wt% Nb

Test Temperature	Material Cold Work	Condition Hydrides	Slot Length, in.	Hoop Strength, ksi	K _c (Calculated), ksi √in.	Failure Mode
Room Temp.	30%	None	None	114	---	Ductile
Room Temp.	30%	None	1/2	94	294	Ductile
Room Temp.	30%	None	1 1/2	42	238	Ductile
Room Temp.	30%	None	2 1/2	25	293	Ductile
Room Temp.	30%	None	4	26	619	Tension
Room Temp.	60%	None	None	121	---	Ductile
Room Temp.	60%	None	1/2	92	236	Ductile
Room Temp.	60%	None	1 1/2	41	229	Ductile
Room Temp.	60%	None	2 1/2	21	244	Ductile
Room Temp.	60%	None	4	22	516	Ductile
0 °C	30%	None	None	115	---	Tension
0 °C	30%	None	1/2	95	294	Ductile
0 °C	30%	None	1 1/2	43	251	Ductile
0 °C	30%	None	2 1/2	26.6	300	Ductile
0 °C	30%	None	4	28	669	Ductile
0 °C	60%	None	1 1/2	43	---	Ductile
-30 °C	30%	None	None	118	---	Tension
-30 °C	30%	None	1/2	92	245	Ductile
-30 °C	30%	None	1 1/2	43.5	246	Ductile
-30 °C	30%	None	2 1/2	29	343	Ductile
-30 °C	30%	None	4	32.5	785	Ductile
-30 °C	60%	None	None	127	---	Tension
-30 °C	60%	None	1/2	95	241	Tension
-30 °C	60%	None	1 1/2	40.5	220	Tension
-30 °C	60%	None	2 1/2	24	270	Tension
-30 °C	60%	None	4	14.2	329	Tension
150 °C	30%	None	None	85	---	Ductile
Room Temp.	60%	None	None	99.5	---	Ductile
300 °C	30%	None	None	84		Ductile
300 °C	30%	None	2 1/2	15.6	179	No Propagation
300 °C	60%	None	1/2	64	---	No Propagation
300 °C	60%	None	2 1/2	16	---	No Propagation
300 °C	30%	275 ppm H ₂	1/2	73	---	Tension
300 °C	30%	275 ppm H ₂	1 1/2	30	---	Tension
300 °C	30%	275 ppm H ₂	2 1/2	27	---	Tension
300 °C	60%	275 ppm H ₂	1/2	85	---	Tension
300 °C	60%	275 ppm H ₂	1 1/2	8.8	---	Tension
300 °C	60%	275 ppm H ₂	2 1/2	7.9	---	Tension
~50 °C	30%	275 ppm H ₂	1/2	26	---	Tension
150 °C	30%	275 ppm	1 1/2	26.3	---	Tension
150 °C	60%	275 ppm	1/2	34	---	Tension
300 °C	30%	275 ppm	1 1/2	34	---	No Propagation
300 °C	30%	275 ppm	2 1/2	8.2	---	No Propagation
300 °C	60%	275 ppm	None	87	---	Tension
300 °C	60%	275 ppm	1/2	43.5	84	No Propagation

$$K^2 \text{ (Calculated)} = \frac{\pi \sigma_f^2 C}{\cos \theta} \left(1 + \frac{5 \pi \lambda^2}{32} \right) \left(\frac{4 - \kappa}{2} \right)$$

where

- κ = (3 - 4μ)/(1 + μ) (Plane Stress Conditions),
- μ = POISSON'S RATIO,
- C = 1/2 SLOT LENGTH,
- R = TUBE RADIUS,
- t = WALL THICKNESS,

σ_f = NOMINAL HOOP STRENGTH OF SLOTTED TUBE,
σ = NOMINAL HOOP STRENGTH OF UNSLOTTED TUBE,

$$\theta = \frac{\pi}{2} \frac{\sigma_f}{\sigma},$$

σ_f = NOMINAL HOOP STRENGTH OF SLOTTED TUBE,
σ = NOMINAL HOOP STRENGTH OF UNSLOTTED TUBE,

$$\theta = \frac{\pi}{2} \frac{\sigma_f}{\sigma},$$

$$\lambda^2 = \frac{C^2}{Rt} \sqrt{12(1 - \mu^2)}.$$

TABLE II. Crack Propagation Tests of Heat Treated Zr-2.5 wt% Nb Tubing

Specimen I Identity	Test Temperature	Material Heat Treatment	Condition Hydride	Slot Length, in.	Hoop Strength, ksi	$K_{IC}^{(1)}$ ksi $\sqrt{\text{in.}}$	Fracture Mode
18A	23	NotAged	None	None	126 ^(a)	---	Shear
14A	23	NotAged	None	None	126	---	Shear
29A	23	NotAged	None	1/2	92	197	Shear
16A	23	NotAged	None	1 1/2	33	147	Tension
12A	23	NotAged	None	2 1/2	28	240	Tension
28A	23	NotAged	None	4	25	422	Tension
20	23	Aged	None	None	122	---	Shear
29	23	Aged	None	1/2	82	160	Tension
16	23	Aged	None	1 1/2	38	175	Tension
12	23	Aged	None	2 1/2	29	250	Tension
10	23	Aged	None	4	13	215	Tension
19A	23	Aged	~275 ppm	None	100	---	Tension
		Aged	~275 ppm	1/2	39	71	Tension
		Aged	~275 ppm	1 1/2	10.8	57	Tension
		Aged	~275 ppm	2 1/2	4.2	44	Tension
		Aged	~275 ppm	4	16.8	393	Tension
13A	150 °C	NotAged	None	None	105	---	Shear

(a) This tubular specimen was tested with a Zircaloy-2 liner having a 0.030 in. wall thickness.

FIGURE 2. Crack Propagation Behavior of Zr-2.5 wt% Nb Alloy Tubing in the 30% Cold-Worked Condition

FIGURE 3. Crack Propagation Behavior of Zr-2.5 wt% Nb Alloy Tubing in the 60% Cold-Worked Condition

FIGURE 4. Crack Propagation Behavior of Zr-2.5 wt% Nb Alloy Tubing in the Heat-Treated Condition

and the heat treated material does not appear to be directly attributable to hydrogen alone. Material anisotropy differences may be a significant factor.

For all the milled slot tests, the fracture strengths for the 4-in. slots appear to be less predictable than for the shorter slot lengths. For the 30% CW test specimens, the fracture strength for a 4 in. long slot is greater than for shorter slot tests. Re-examining the calculated K values as tabulated in Tables I and II, one finds, with a few exceptions, that K_c is essentially constant for the 1/2, 1-1/2 and 2-1/2 in. long milled slots. For the 4-in. long slots--roughly twice the tube diameter--calculated K_c values are generally about twice as great. The reasons for the difference is not obvious but may be partially due to bending forces arising from the internal pressure in the tubular test specimens.

In Zircaloy-2 tubing, it was almost always found that at room temperature, crack propagation immediately followed failure of the web under the milled slot; whereas, in 30% and 60% CW Zr-2.5 wt% Nb tubing, the crack, resulting from failure of the web under the milled slot, was not visible at the outer surface.

Both theory and experiments on flat sheet have shown that in general the fracture stress is proportional to the square root of the crack length.⁽²⁾ For the milled slot tests, the slope of log fracture strength versus log crack length is about 0.7 and appreciably higher than the slope predicted by theory.

A comparison of the fracture strength versus slot length of Zr-2.5 wt% Nb and Zircaloy-2 as presented graphically in Figure 5, shows that there is no major difference between the two alloys. At the shorter slot length, the 2.5 wt% Nb may be slightly more crack resistant. For a slot 1 in. or longer, the fracture strength of unhydrided Zr-2.5 wt% Nb alloy and

FIGURE 5. Comparison of the Crack Propagation Characteristics of Zircaloy-2 and Zr-2.5 wt% Nb Alloys

Zircaloy-2 in both irradiated and nonirradiated conditions are indistinguishable. As previously mentioned, the effect of hydriding (250 to 300) is not readily predictable. For Zircaloy-2 there is one data point for the hydrided and irradiated condition. For this one test, the fracture strength has been appreciably reduced. As yet, one could not determine whether this reduction is due to combined irradiation and hydriding or hydriding alone.

The fracture strength versus crack length curves for test material which has been uniformly hydrided to the saturation level at about 400 °C (e.g., 250 to 300 ppm H₂) are also shown in Figures 2, 3, and 4. For the 30% CW material, the effect is significant but not nearly so drastic as in the case of the 60% CW and heat treated test materials. In these latter materials, hydriding appears to have its greatest effect on 1-1/2 and 2-1/2 in. slots (Table I). For any of the test materials hydrided to a 250 to 300 ppm concentration, the crack propagation behavior could not be reliably predicted. There appears to be little doubt that with flaws present, hydriding to 200 to 300 ppm levels would tend to limit the serviceability.

By use of the fracture data obtained, the stress intensification factor K was calculated. Calculations were based on the relationship between failure stress and crack length for a through-the-wall crack as formulated by T. J. Atterbury, et al.⁽³⁾ This relationship was developed for a longitudinal through-the-wall crack in the cylinder under internal pressure.

$$K^2 = \frac{\pi c \sigma_f^2}{\cos \theta} \left(1 + \frac{5\pi \lambda^2}{32} \right) \frac{(4 - k)}{2}$$

where

K = stress intensity factor - ksi√in.,

c = one-half of the length of the through-the-wall crack (inches),

σ_f = fracture hoop stress for slotted tube (ksi),

σ = fracture hoop stress for unslotted tube (ksi),

$$\theta = \frac{\pi}{2} \frac{(\sigma_f)}{\sigma},$$

$k = (3 - 4\nu)$ for plane strain and

$$\frac{(3 - \nu)}{(1 - \nu)} \text{ for plane stress,}$$

ν = Poisson's Ratio,

$$\lambda^2 = \frac{c^2}{Rt} \sqrt{12 (1 - \nu^2)}$$

R = radius of the tubular test specimens (inches),

t = tube wall thickness (inches).

It was assumed that plane stress conditions more nearly approximated the stress state in the tubular test specimens, although it is likely that in all tests some radial constraint was present at the crack tip. The calculated K values at fracture for the milled slot test specimens are in Tables I and II. For all three test materials, the lowest calculated K values are obtained for the 1 1/2 in. long slot. For the 4 in. long slot, the calculated K values are large and roughly twice those obtained for the 1/2, 1-1/2, and 2-1/2 in. slots. For the unhydrided material, calculated K values are significantly lower for the heat treated test materials. Calculated K values obtained from tests on hydrided test material, except for the 4 in. long slot, are significantly lower than those from any of the unhydrided test material. Those test materials, heat treated and the hydrided and heat treated, in which the crack tended to propagate immediately upon failure of the remaining web of the milled slot, are also the materials from which the lowest K values are obtained.

Fatigue Crack Tests

The calculated values of the stress intensification factor at fracture seemed quite high as compared to those

obtained from single edge notched and notched tensile tests.⁽⁴⁾ Because of the large K values calculated from milled slot test data, a few fatigue tests were done. These tests were low stress-low cycle tests to produce pseudo-natural cracks in tubular fracture test specimens. As described previously, these fatigue tests employed a peak hoop stress of about 23,000 psi tension and a minimum hoop stress of about 1000 psi tension, a cycle rate of about 400 cycles/hr, and small electrical discharge machined slots (EDM slots) as stress risers. A test with an EDM slot, semicircular in profile and a radius of 0.125 in. and a width of 0.010 in. or less, resulted in fatigue cracking at only a few thousand stress cycles, i.e., fatigue cracking was observed on the surface at less than 6000 cycles (Figure 6). At 9700 cycles, leakage occurred with a crack length of slightly more than 0.4 in. as observed on the outside surface of the test specimen. To verify that leakage would occur at the same crack length, three additional tests were done with the same cyclic stress loading and cycle rates. In these tests, the semicircular EDM slot shape was retained. The radius of the EDM slot, however, was progressively increased to provide EDM slot lengths of 0.10, 0.15, and 0.20 in. at the outer surface of the tubular test specimens. Decreasing the slot length would increase the number of stress cycles needed to initiate and grow a fatigue crack to the critical length for leakage if the leakage crack were only a function of test specimen geometry. In all four tests, leakage, at the peak cyclic stress, did occur at a fatigue crack length at the outer surface of about 0.4 in. and confirmed that the leakage crack length is only dependent on test specimen geometry (Figure 6). After the leakage crack length is reached, the number of cycles needed to grow the fatigue crack to critical length for the same peak cyclic hoop stress appears to be about the same or decrease slightly with increasing amounts of fatigue damage. The critical crack length for catastrophic failure diminishes as the number

FIGURE 6. Fatigue Crack Length as a Function of Fatigue Loading Cycles

of cycles required to grow the crack increases. The results of these four tests (Specimen No. 9, 11, 22, and 22-A) indicate that the critical crack length is to some degree at least, a function of prior fatigue damage. As the total amount of fatigue damage increases, prior to failure, the critical crack length diminishes for failure under the same cyclic stress conditions. Based on this hypothesis, it would appear to be possible to incur catastrophic failure at or before the crack grows to leakage dimensions providing that the initial flaw size is sufficiently small. Referring to Figure 6, the number of cycles for such an occurrence is estimated to be well above 100,000 cycles, or well above the number of cycles a pressure tube would receive in service.

Several additional fatigue tests were done which are of interest. The results of these tests are shown graphically in Figure 7 and listed in Table III. In the first (Specimen No. 24), a spot weld was used as the stress riser rather than a mechanical defect (e.g., EDM slot). The spot welder power setting was high enough to raise the metal temperature locally into the orange color range. A small tab of zirconium hydride (stoichiometric ratio of zirconium and H_2) was welded onto the outside surface in an effort to drive H_2 locally into the tubular test specimen. Based on metallographic examination of the surfaces after failure, the hydrogen content at the outside surface was estimated to be several hundred ppm; little H_2 diffused through the tube wall to the inner surface. Even though little hydriding was produced, the local heating and rapid cooling was believed to be sufficient to introduce high local stresses which could initiate fatigue cracking. Fatigue cracking did occur, but about 62,000 cycles were required to initiate cracking visible at the outside surface. Once, cracking was observed at the outside surface; crack growth was rapid in the heat affected area and failure occurred in

FIGURE 7. Fatigue Crack Lengths Versus Number of Fatigue Loading Cycles For Several Special Tests

TABLE III. Fatigue Crack Propagation Tests of Heat Treated Zr-2.5 wt% Nb

Specimen Identity	Test Temperature	Material Heat Treatment	Condition Hydride	Crack Length, in.	Hoop Strength, ksi	$K_C \sqrt{\text{in.}}$	Fracture Mode
26A	23	NotAged	None	1 1/2	21	111	---
30-11 SR	23	(a)	None	OD-0.68 ID-0.43	58	148/102	Shear
	23	Aged	None	1.48	21	114	Tension
6	23	NotAged	None	1.36	23	114	Tension
24 ^(b)	23	NotAged	(b)	1.32	24	113	Tension
11	23	Aged	None	1.31	23	112	Tension
9	23	NotAged	None	1.36	23	114	Tension
22	23	Aged	None	1.08	23	84	Tension
27 ^(c)	23	Aged	None	>2	20	153	Tension
22A	23	Aged	None	0.95	23	73	Tension

- (a) This test specimen was in the 30% CW condition and had been subjected to a creep test for about 12,000 hr at 300 °C. Total creep strain was about 2%.
- (b) The stress concentrator consisted of a spot weld and local hydriding. Local hydrogen concentration was estimated to be about 100 ppm.
- (c) The stress concentrator consisted of an EDM slot with a radius of 0.125 in. located on the ID of the test specimen.

relatively few fatigue cycles. The test results are shown in Figure 7 and Table III.

In another test, a test specimen (Specimen No. 26A), having a 1 1/2 in. long and 1/16 in. wide milled slot with a depth equal to 80% of the tube wall thickness, was fatigue cycled after the web remaining from the milled slot had failed. Abrupt catastrophic failure occurred after only 1550 cycles. Metallographic examinations showed that the crack formed from the failure of the web remaining from the milled slot was extended only a few thousands of an inch by fatigue before catastrophic failure occurred. The test results are shown in Table III.

In a similar specimen (Specimen No. 2) with a 1/2 in. long milled slot, fatigue cracking could not be induced at a similar hoop stress in nearly 45,000 cycles. At this point an EDM slot, 0.180 in. long, less than 0.010 in. wide, and with a depth equal to 80% of the tube wall, was cut from one end of the original milled slot. Fatigue cycling now caused fatigue cracking from one end of the EDM slot and catastrophic failure occurred when the total combined milled and EDM slot and fatigue crack lengths reached about 1.48 in. (Table III). While it seems likely that catastrophic failure was initiated at the fatigue cracked end of the defects, it was found that catastrophic crack propagation also occurred from the milled slot end of the defect.

In one test (Specimen No. 27), the EDM stress riser slot was located on the inside surface of the test specimen. Leakage occurred at a very short crack length as measured on the outside surface (Figure 7). The crack length on the inside surface must have been much longer since the outside surface crack length was about one-half the original EDM slot length at the inside surface. Based on the so-called fatigue crack growth laws,⁽⁵⁾ the number of cycles to leakage

was appreciably greater than expected--using the results obtained from an earlier test (Specimen No. 9, Figure 6)--with the same size stress riser EDM slot located at the outer surface. When the wide variability of results often found in fatigue life tests is considered, this difference may not be significant. One factor appearing to be significant is that the critical crack length at catastrophic failure is longer than for the other tests. The resulting calculated K value is, as a consequence, significantly larger than for the other tests (Table III). The reason for the observed differences are not now understood.

One other test utilizing an EDM slot on the outer surface resulted in a significantly different calculated K value. The test specimen (No. 30-11-SR) involved was from the 30% cold-worked lot of tubing; it had exhibited 2% creep strain during a creep test, stressed by internal pressure, for nearly 12,000 hr at 300 °C. After about 6300 cycles, the fatigue crack had grown to a length of about 0.70 in. as measured on the outside surface. It was then statically pressurized until failure. Postfailure examination showed that the fatigue crack- ing extended uniformly from the 0.150 in. radius EDM slot, and it retained the semicircular profile. Hence, the fatigue crack length, as measured on the inside surface, was 0.43 in. Depending upon whether the inner or outer surface crack length is used, calculated K values are 102 and 148, respectively. By using an average of these two, the K value is then within reasonable agreement with most of the other tests (Table III).

In Figure 8, the logarithm of the ratio of "flawed" to "unflawed" hoop strength is plotted against the logarithm of fatigue crack length. Most of the tests were run at a peak cyclic hoop stress of about 23,000 psi. Little data was obtained for crack lengths of less than 1 in. or greater than 2 in. Test 30-11-SR described above was terminated when the

FIGURE 8. Fatigue Crack Propagation Behavior of Zr-2.5 wt% Nb Alloy Tubing in the Heat-Treated Condition

fatigue crack was about 0.70 in. By averaging the inner surface (0.43 in.) and outer surface (0.70 in.) crack lengths, the resulting curve shows a slope of about unity. Again, the slope of this curve is appreciably greater than predicted by brittle fracture theories. However, the hypothesis (as stated earlier) that the critical crack length decreases with increasing fatigue damage suggests that there may be no unique curve or relationship between failure stress and crack length.

In Figure 9, the fracture strength curves are shown for milled slot and fatigue crack type of tests. It is obvious that, in the 1 to 2 in. crack length range, a fatigue cracked tubular test specimen fails catastrophically at a much lower hoop stress. This fact is also apparent in the calculated K_c values (Tables I, II and III). The magnitude of the K_c values for milled slot tests are at least twice those obtained from fatigue tests. From these observations, one tentatively concludes that for the fatigue crack the stress concentration is greater than for a milled slot; that the amount of plastic deformation is greater when a milled slot begins to fail; and fatigue damage appears to increase the materials susceptible to "brittle" fracture behavior.

CONTINUING WORK

Fatigue tests have shown that both fatigue crack growth and brittle fracture data can be readily obtained from the same tests. Some of the fatigue data obtained suggest that the critical crack length for catastrophic failure may be influenced by prior fatigue damage. In these same tests, the leakage crack length appeared to be only a function of geometry and stress. Presumably, if enough fatigue damage could be accumulated, catastrophic failure might be influenced by prior fatigue damage. For some applications a "leak-before-failure" service criteria is practical. However, one needs to establish the conditions and extent to which such a service criteria can

FIGURE 9. Crack Propagation Behavior of Zr-2.5 wt% Nb Alloy in the Heat-Treated Conditions Using Milled and Fatigue Produced Flaws

be safely and economically applied. Hence, some further testing of small short "defects" of the EDM slot type appears to be warranted.

Some preliminary work has been done in fatigue testing with "defects" in close proximity. It appears that further experimental and analytical evaluation of the interaction of multiple cracks is needed. These investigations could be practical because not only is flaw size important, but relative proximity of one flaw to another could well determine whether catastrophic failure might occur.

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