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The Effects of Coupling Nuclear Radiation with Static and Cyclic Service Stresses and of Periodic Proof Testing on Pressure Vessel Material Behavior

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ABSTRACT

The nuclear service performance of structural steels as influenced by static and cyclic stress applications during radiation exposure has been examined and documented with experimental results. The significance and merits of initial and subsequent proof tests of large structural components such as the hydrotesting of nuclear reactor pressure vessels have also been reviewed and evaluated. Performance following preload in the form of warm prestressing as well as aging embrittlement were among those factors considered.

A summary review of available data representing a coupling of static stress and radiation has suggested neither a significant effect of stress on irradiation responses, nor a synergistic effect of stress and radiation on mechanical properties performance. New data also indicate that stress during irradiation may not have a general influence on post-irradiation annealing response. A parallel review of the cyclic stress behavior of pressure vessel steels did reveal a slight alteration of basic (preirradiation) low-cycle fatigue properties by neutron exposure. The general trend depicted was one of a radiation reduction of fatigue life at high strain amplitudes (low-cycle failure region) and of improvement at low strain amplitudes (high-cycle region). The trend was found consistent for both pre-post irradiation test methods and dynamic in-reactor testing.

Evaluations of the merits and objections for proof testing allowed a conclusion that both initial and periodic proof tests conducted above the nil-ductility transition temperature (NDT) are of definite value; however, operations below NDT either as a proof test or subsequent to warm prestressing above NDT were found objectionable.

PROBLEM STATUS

This is a final report on one phase of the problem; work on other phases is continuing.

AUTHORIZATION

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THE EFFECTS OF COUPLING NUCLEAR RADIATION WITH STATIC AND CYCLIC SERVICE STRESSES AND OF PERIODIC PROOF TESTING ON PRESSURE VESSEL MATERIAL BEHAVIOR

INTRODUCTION

Knowledge concerning the effects of nuclear radiation upon the physical and mechanical properties of structural steels and alloys has grown immensely within recent years. As one result of this growth, comprehensive data correlations and more consistent trends in property changes as functions of neutron fluence levels and service temperature are becoming available for reactor design and operation. At the same time, however, the interrelating of irradiation responses of selected alloys to that of a single "reference" alloy, as well as correlations of the effects of one environment versus another, have revealed in some cases significant and as yet unexplained material-performance deviations. For example, higher-strength steels were noted to depict a general trend of decreasing radiation embrittlement sensitivity with increasing (preservice) yield strength level (1). However, 5Ni-Cr-Mo-V steel (130/140 ksi YS) proved to be a major exception (2). Furthermore, it has been shown that the neutron spectral characteristics of light-water-moderated test reactors permit the interrelation and comparison of radiation effects on structural steels through the use of a fairly common method of assessing neutron fluence (e.g., $n/cm^2 > 1$ Mev, fission) (3). By contrast, data from heavy-water-moderated reactor irradiations cannot be successfully interrelated to alloy performance in a light-water-moderated environment using the same rules or dosimetry techniques (4).

The observation of experimental anomalies, including variations from mechanical-property trends and differences in degree of environmental influences, leads to concern for potential anomalies in other areas. One such area of concern is that of synergistic effects of environment and stress, while another is that of a possible compounding of known effects of stress by radiation exposure (e.g., strain aging). It is to the uncertainties of the effects of coupling nuclear radiation with static and cyclic service stresses that Parts 1 and 2 of this report are addressed. Part 3 reviews the effects and merits of periodic proof testing as a conventional demonstration test, including alloy behavior following prestressing. New and other unpublished experimental data are examined in concert with data reported more recently in each of three separate areas: the effects of static stress in radiation applications, the effects of irradiation on fatigue cycling response, and the effects of prestress and proof testing on materials serviceability. It is shown that studies involving static and cyclic stress applications during irradiation are still largely of an exploratory nature due to a "cause and effect" situation with regard to the degree of difficulty in the successful design and performance of irradiation-stress experiments, relative to conventional specimen (unstressed) exposures. Similarly, it is shown that the effects of periodic proof testing have not been fully explored or qualified in depth, leading to a measured degree of uncertainty in this area.

For simplicity of presentation, the three areas of investigation as outlined are treated separately.

Part 1

EFFECT OF A LARGE STATIC STRESS ON IRRADIATION RESPONSE OF PRESSURE VESSEL STEELS

INITIAL RESEARCH AND ALTERNATE EXPERIMENTAL APPROACHES

Among initial experimental attempts to qualify the possible effects of a superimposed stress upon the radiation behavior and performance of carbon and low alloy steels were the exploratory studies of Nichols and Harries (5) and of Reynolds (6). Both of these investigations, as well as more recent investigations by NRL, selected notch ductility properties for study as being a highly sensitive indicator of any changes in irradiation response.

The work by Nichols and Harries, involving aluminum-grain-size-controlled and silicon-killed pressure vessel steels, illustrates one approach that has been taken to examine the effects of stress and irradiation. Charpy-V specimens from unstrained and two-percent prestrained test sections of the two steels were irradiated together at temperatures in the range of 150° to 300°C (302° to 572°F) to neutron fluences up to 2×10^{19} n/cm^2 (fission).* On comparing their postirradiation test results for both steel conditions, these investigators concluded that "2 percent strain has no effect on the irradiation embrittlement of pressure vessel steels over a wide range of irradiation temperature and dose conditions (5)." While their observations were received with great interest by the research community, they recognized that the "two-step approach," i.e., straining followed by irradiation in the unstressed state, may not have provided the critical conditions for revealing synergistic effects as possibly would the "one-step approach" involving a simultaneous application of both. In particular, experimental provisions for this combination were felt to be necessary, since neutron-induced damage is generally associated with dislocation production which, furthermore, may be partially offset by a self-annealing process at some reactor service temperatures.

Recognizing the ultimate need for simultaneously stressing and irradiating structural materials, Reynolds pursued the "one-step approach" through the design and development of a rather unique irradiation assembly for use in the Vallecitos Boiling Water Reactor (VBWR). With his design (6), a single long bar having a 0.394-in. square cross section (equal to that of a Charpy-V specimen) could be irradiated under stress, along with a second, unstressed bar for direct comparison tests. Each bar was sufficiently long for later sectioning into 10 to 12 Charpy specimens for postirradiation assessments. By design, the irradiation assembly was open to the reactor coolant so that the combination of coolant pressure (about 1000 psig) acting upon evacuated bellows could be used with suitable linkages to apply a static load to the test bar. With this arrangement, a tensile stress on the order of 13.5 ksi could be developed under normal reactor operations. Unfortunately, this stress level was the maximum for the cross-sectional area of the test bar and was set by the basic design and the (space) limitation on allowable bellows size. In terms of preirradiation tensile properties of the A302-B plate investigated by Reynolds, the applied load was ~19 percent of the yield stress and 14.7 percent of the

*Termed as "fast" fluence, or the neutron fluence calculated assuming a fission spectrum which would produce the observed activity in the flux monitor (5).

ultimate tensile strength. Therefore, the experimental conditions for the intended assessment of the effects of simultaneous irradiation and stressing were not too severe. Reynolds' open assembly design also required that the specimens be protected from the corrosive effects of the reactor coolant. A thin chromium plate was applied to both reference and control test specimens, followed by a "bake-out" at 350°F, presumably for hydrogen removal. What effect, if any, this process may have on the test data is uncertain. It is not the primary purpose of the foregoing discussion to point out specific experimental or design problems in this particular investigation, but to illustrate clearly the difficulties encountered in truly qualifying the combined effects of irradiation and stress.

The data reported for the single experiment performed with the VBWR assembly was analyzed by Reynolds as showing somewhat greater impact strength for the stressed material, in spite of a somewhat lower exposure on the unstressed bar (due to reactor flux gradients). In explanation, Reynolds suggested that "stress during irradiation possibly aids the growth of vacancy clusters beyond the size and spacing at which the maximum effect on mechanical properties occurs (6)." Thus, the question of a possible enhancement of separate aging phenomena was introduced.

To synopsize the early exploratory efforts of Nichols and Harries and of Reynolds, two different approaches to the examination of stress effects are represented. Each study had certain limitations, either real or implied, as a result of which the engineering concern for an enhancement of irradiation effects through stress application during exposure could not be either dismissed or fully qualified. The results of the two independent investigations were not in general accord; the direct approach taken by Reynolds implies an effect of stress, while the indirect approach representing the United Kingdom (UKAEA) study proved negative. To a large measure, the research effort by NRL in this area was initiated in the hope of resolving the experimental uncertainties. Its primary purpose, however, was to allow the critical assessment of the relevance to reactor design of the volume of data being generated by NRL and other laboratories with specimens exposed in the unstressed condition (1) in various material test reactors.

NRL STUDIES ON STATIC STRESS DURING NEUTRON IRRADIATION

Irradiation Assembly Design and Experimental Techniques

NRL studies of the combined effects of stress and radiation reflect the "one-step approach" and, seeking to improve upon earlier research, were aimed at providing high static loading conditions during irradiation. Specifically, loading conditions were to be maintained at or in excess of 75 to 80 percent of the initial yield stress of the steel under investigation. Program requirements included the simultaneous irradiation of a sufficient number of specimens to develop complete ductile-brittle transition curves for each steel condition. It was also planned that specimen exposures would be in a helium-air environment to eliminate the need for special corrosion protection, and that specimen temperatures would be controlled at preselected levels representative of reactor pressure vessel service. Each of the above requirements was ultimately satisfied by the final in-reactor exposure assembly design, in conjunction with special experimental techniques and procedures to be described. It will be noted that, commensurate with the intended exploratory study, simplicity rather than the extremes of experimental precision received the greater emphasis in assembly-development phases. Design limitations resulting from this emphasis will be brought out in ensuing discussions of the various system features.

The experimental assembly developed by NRL for the simultaneous exposure of stressed and unstressed reference specimens for subsequent Charpy-V determinations is shown in Fig. 1. The assembly employs five bars of multiple-specimen length, as shown in Fig. 2. The square, reduced section of the bars conforms to the cross section

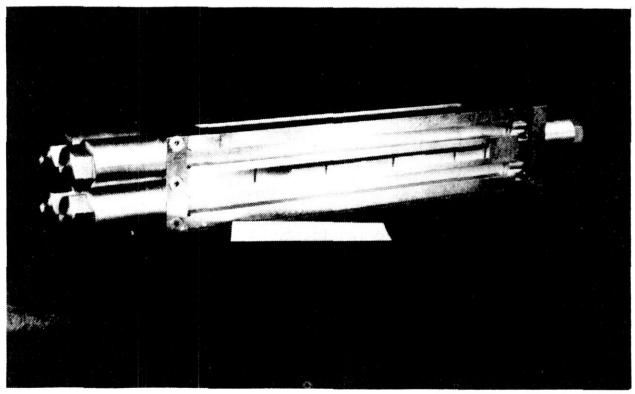

Fig. 1 - Experimental assembly for the simultaneous irradiation of specimens under stress and Charpy-V reference specimens for studies of notch ductility behavior

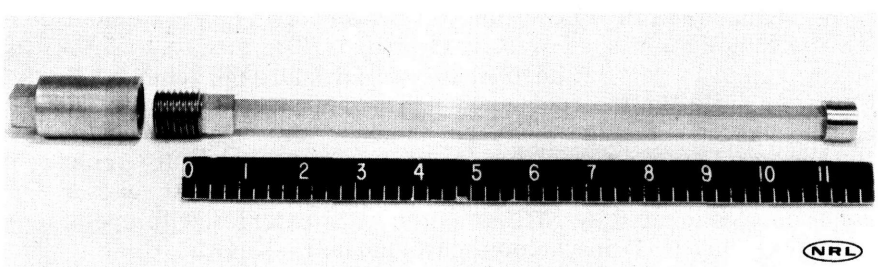

Fig. 2 - Design of specimen stressed during irradiation. Reduced section (0.394 × 0.394 in.) conforms to the cross section of a standard Charpy-V specimen. Square shoulder adjacent to threaded end serves as a rotation lock while the specimen is being stressed.

of a standard Charpy-V specimen (0.394 × 0.394 in.) and is of sufficient length to provide four specimens per bar, or a total of 20 specimens per assembly. Unlike Reynolds' design, it is noted that the NRL specimen assembly is relatively short to allow the positioning of the specimens within the "flat" flux region of the irradiation exposure facility. Furthermore, the bar (Fig. 2) is stressed by means of a takeup nut on its threaded end, the square shoulder at that end serving as a rotation lock to prevent torsion loading. Strain gages placed at several locations over the length of each bar are used to monitor axial loading conditions. Thus, maximum load application is an experimental option and is not grossly limited by design. As will be discussed in a later section, strain measurements are made both before and after irradiation as a standard practice to insure that reported strain values were maintained over the entire period of irradiation exposure. As an aid to this objective, from the standpoint of thermal expansion, it was planned that whenever possible the body structure of the assembly would be of the same parent stock as the test specimens, so that the bars would not receive (or lose) additional strain from differential thermal expansion.

The assembly body shown in various stages of fabrication in Fig. 3 also contains slots for 20 unstressed reference specimens. The final assembly is double encapsulated for temperature control by the internal pressurization system previously described (7), for

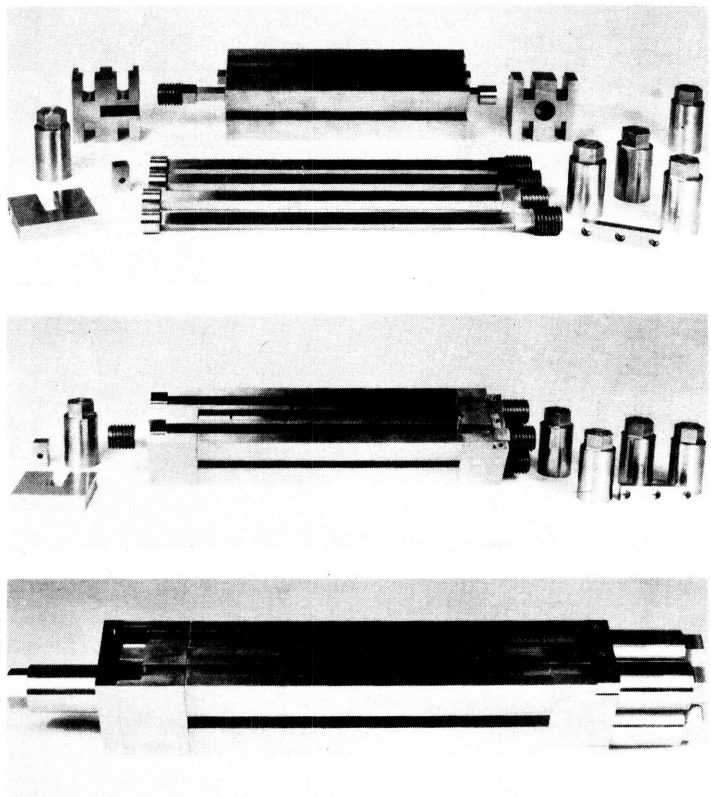

Fig. 3 - Experimental assembly for the simultaneous irradiation of stressed and unstressed (reference) specimens, showing component design and unit construction at various stages

dynamic in-reactor fatigue test experiments (Figs. 4 and 5). In brief, the functioning of the system depends upon: (a) the attainment of a static temperature (no external control) developed by gamma heating below that desired for irradiation, (b) a flexible "outer" containment sheath which can be moved away from the enclosed assembly by internal pressurization, and, (c) instrumentation to increase the internal pressure on the sheath automatically when temperatures are too low and to decrease this pressure when temperatures are too high.

In the present application, the internal pressurization system, while providing ±5°F control about the set point, must rely on other design features to minimize temperature variations over the specimen length. Since a longer specimen length is involved here than in the fatigue test assembly, a greater dependence is placed upon axial heat conduction and the helium sweep through the assembly during operation. With the system as designed, it was found that gradients could be held to less than ±15°F at 430°F and to ±20°F at 550°F. Gradients of such magnitude were not felt overly critical to postirradiation assessments; however, the assembly would have an operational limitation if higher, more severe gradients develop through gross gamma heating variations over the assembly length. One other operational limitation to be pointed out involves the inability to correct for a high degree of stress relaxation of the sample material if this event should occur during irradiation. However, it is recalled that strain measurements are also made after irradiation, wherein the degree of stress relaxation can be assessed and accounted for.

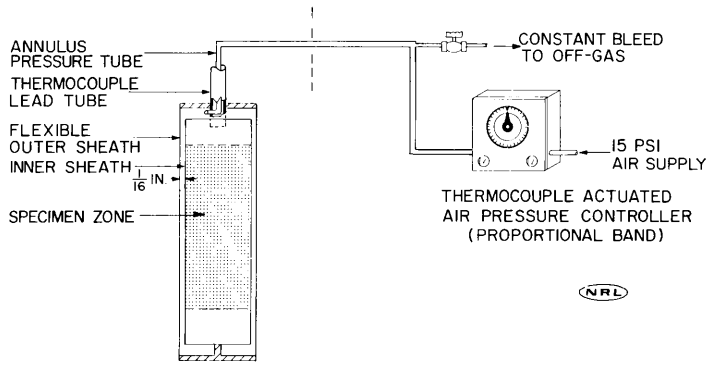

Fig. 4 - General design of the irradiation assembly double containment and air pressurization system for specimen temperature control

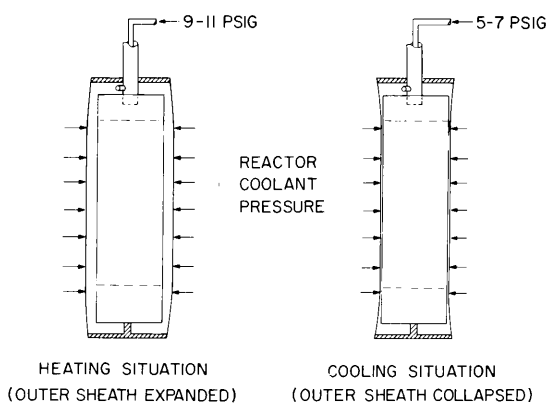

Fig. 5 - Action of the assembly outer containment sheath during specimen heating and cooling situations

Assessments of Two Reactor Pressure Vessel Steels Following Irradiation Under Stress

To date, two independent assessments of the effects of stress on notch ductility behavior during irradiation have been performed by NRL using the special exposure assembly described.

Investigation 1 – A350-LF1 (Modified) Steel – The initial irradiation experiment performed by NRL to assess the effects of irradiation under static stress was sponsored by the Army Nuclear Power Field Office and was directed specifically to the Army SM-1A reactor. This power reactor operates at a nominal temperature of 430°F and has an A350-LF1 (Modified) steel ring forging forming its pressure vessel. Specimens for the investigation were taken from a duplicate ring forging (Table 1) from the same steel heat as the vessel; however, the forging thickness was not sufficient to permit the fabrication of the body of the exposure assembly, and a mild carbon steel was used as a substitute material. The irradiation experiment was performed in a fuel core position of the Oak Ridge LITR and received a neutron fluence of about 3×10^{19} n/cm^2 > 1 Mev.

The minimum strain level desired during irradiation (1300 μin./in.) corresponded to 80 percent of the 430°F yield strength (0.2 percent offset) of the test forging. The nominal strain histories of the stressed bars illustrated in Fig. 6 confirm that this objective was achieved. Preirradiation strain measurements with etch foil strain gages

Table 1
Chemical Composition and Heat Treatment of 2-5/8-in. A350-LF1 (Modified) Steel Ring Forging

Analysis Source	Chemistry (Wt %)										
	C	Mn	P	S	Si	Ni	Cr	Mo	Cu	V	Al
NRL	0.13	0.76	0.023	0.036	0.25	1.64	0.16	0.06	0.08	0.04	0.06

Heat Treatment

Forging: Forged in the temperature range of 1700°F minimum to 2250°F maximum

Post Forging: Normalized at 1600°F, water quenched; Tempered at 1250°F, stress relieved at 1150°F

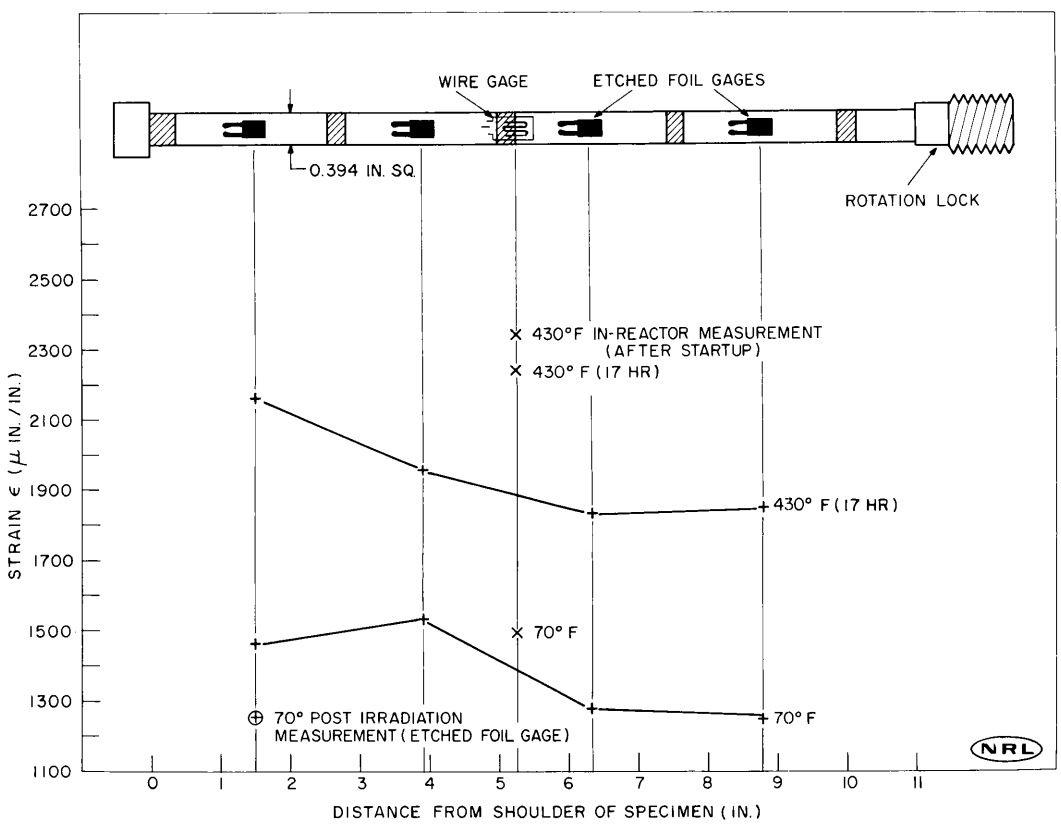

Fig. 6 - Typical strain history of an A350-LF1 (Modified) steel specimen stressed during irradiation

(code +) and with ceramic bonded wire gages (code ×) over the interval 70° to 430°F indicated a progressive rise in strains with an increasing temperature. It has been assumed that this change was due to the body of the stress assembly having a somewhat higher thermal expansion coefficient than the specimen material. Shortly after reactor startup, in-reactor measurements with the wire gages again indicated a rise in strain

with temperature and were in good agreement with preirradiation measurements. It should be noted that while compensating gages were used (external-type), the strain-correction factors for the wire gages at elevated temperatures have not been fully established by the manufacturer. Thus, measurements with the wire gages must be considered as indicating a relative strain condition only. Postirradiation strain measurements, performed remotely as the specimens were unstressed, utilized etched-foil gages only. The application of strain gages in the radiation hot cell is shown in Fig. 7. For simplicity and ease of disassembly, strains of three rather than five bars were monitored during unbolting operations. In the development of this technique, the gamma radiation resistance of the foil gages was assessed by applying duplicate gages to highly radioactive tension test specimens and comparing strain gage response against extensometer readings as the specimens were pulled. As shown in Fig. 6, some stress relaxation was evident; however, the amount of strain loss during irradiation was not sufficient to reduce specimen strains below the desired minimum. Possible causes of an apparent loss in strain will be discussed later.

After removal of the stressed bars from the irradiation assembly, each was cut to the required Charpy specimen length using a remotely operable milling machine (8) and notched with a special broaching fixture. Dimensional control was well within ASTM tolerance specifications for Type A Charpy-V specimens (9).

Experimental results from the 430°F irradiation are shown in Fig. 8. As noted from data points for the unirradiated condition, the material exhibited data scatter characteristic of a forging. However, relative uniformity in notch ductility through the 2-5/8-in. thickness was observed. Such through-thickness uniformity was particularly desirable in this experiment, since the stressed bar, due to its length, traversed several thickness points of the test stock. After irradiation, a Charpy-V 30-ft-lb transition temperature increase of about 400°F, consistent with earlier results on this heat of steel, was observed for the unstressed condition. Significantly, data for the stressed condition fell within the scatter-band of data for the unstressed condition, to indicate an equivalence in irradiation response. It would therefore appear that an applied stress during the 430°F irradiation of this steel neither enhances nor reduces its rate of radiation embrittlement. These results were particularly gratifying, since the experimental conditions were based directly on the SM-1A reactor case.

Investigation 2 – A302-B Steel – The second evaluation of the combined effects of stress and radiation was performed using samples from a heavy section A302-B steel test plate (Table 2). The irradiation exposure was conducted in a fuel-core position of the

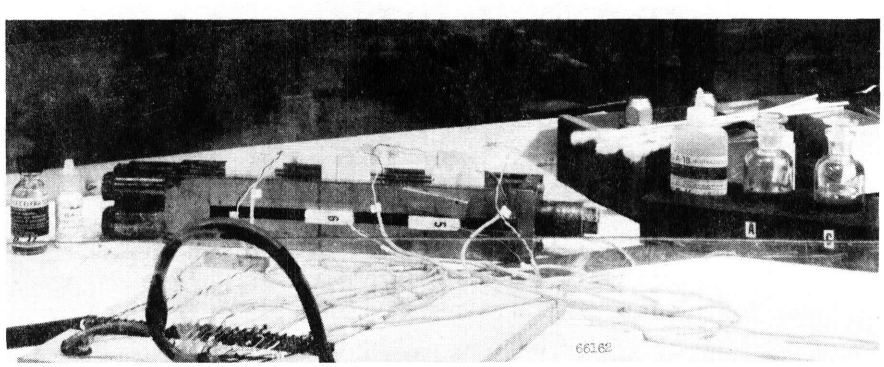

Fig. 7 - View of the irradiated experimental assembly through the hot-cell window, showing the general placement of the strain gage holders on the stressed bars prior to unbolting operations

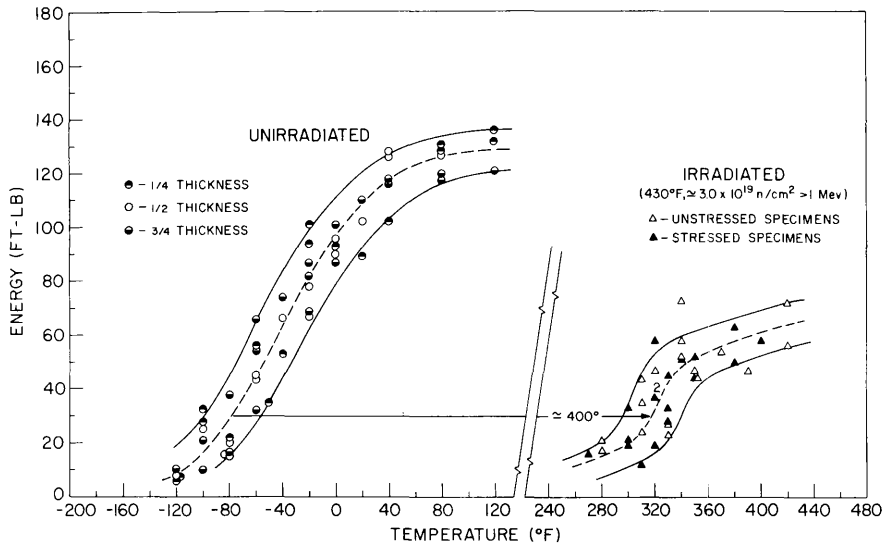

Fig. 8 - Comparison of notch-ductility performance of 2-5/8-in. A350-LF1 (Modified) steel ring forging irradiated in the unstressed and stressed conditions

Table 2
Chemical Composition and Heat Treatment of 10-1/2-in. A302-B Plate

Analysis Source	Chemistry (Wt %)										
	C	Mn	P	S	Si	Ni	Cr	Mo	Cu	V	Al
NRL	0.25	1.37	0.006	0.011	0.21	0.27	0.10	0.62	0.15	≤ 0.01	0.02
WAPD	0.20	1.37	0.009	0.017	0.19	–	–	0.48	–	–	–

Heat Treatment

Austenitized: 1550°-1600°F for four hours, water dip quenched

Tempered: 1225° ±25°F for four hours, furnace cooled

Stress Relieved: 1150 ±25°F for 30 hours, furnace cooled to 600°F

Union Carbide Reactor, Tuxedo, New York, at a controlled temperature of 550°F, chosen as best representing current pressurized and boiling-water-reactor applications of this steel. The neutron fluence received by the assembly was $\simeq 2.0 \times 10^{19}$ n/cm^2 >1 Mev, determined from an evaluation of the dosimetry results of 32 Fe^{54} flux monitors included in the unit.

Patterned after the initial investigation with the A350 steel, the target stress level in this case was also 80 percent of the elevated temperature yield strength. As depicted in Fig. 9, this stress was determined to be 42,500 psi, corresponding to a strain level of about 1650 μin./in. In this plot, the two stress-strain traces given for each of the two

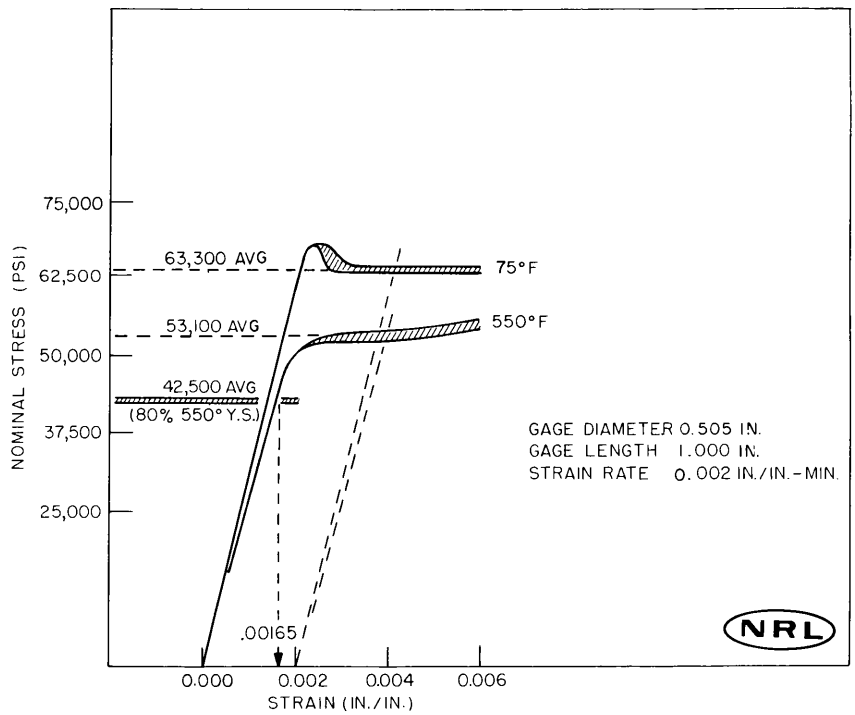

Fig. 9 - Tension test behavior of 10-1/2-in. ASTM A302-B steel plate at ambient temperature and at 550°F. Curves are reproductions of actual load-extension traces for irradiated test specimens.

test temperatures represent maximum and minimum behavior found in multiple specimen testing. Thus, it can be seen that the planned minimum strain level was within the elastic load range of this material.

In this second study, the test stock (10-1/2-in.-thick plate) was sufficient in size to allow the fabrication of the body pieces of the exposure assembly, and an exact match of thermal expansion coefficients of specimens and assembly body proper was therefore possible. Preirradiation tests of assembly performance at 350°, 450°, and 550°F revealed that specimen strain levels were either constant or slightly increased at these elevated temperatures. Because of the higher assembly operating temperature (550°F) in this second study, "Rokide-bonded" wire Type HT 235-4A-S6 (Nichrome) strain gages were used. Stress relaxation was not observed on returning to ambient temperature (75°F) following any one preirradiation test, including a final four-day test at 550°F. However, postirradiation strain measurements made in-cell with foil-type gages as the specimens were unstressed did indicate that some stress relaxation had taken place during the six-week irradiation period. The average postirradiation strain level of the three bars measured was 1600 μin./in., or slightly below the target strain level for this experiment. However, this minimum strain indicated that a stress level in excess of 75 percent of the 550°F preirradiation yield strength (0.2 percent offset) of this steel was maintained throughout the duration of the neutron-exposure period.

Experimental results from Charpy-V tests are presented and compared in Fig. 10. Referring first to the data for the unirradiated test condition and the schematic diagram showing specimen locations within the test plate, relatively uniform notch ductility properties were exhibited by a 1-3/4-in.-thick volume of metal centered on the quarter thickness plane. On the basis of this observed uniformity, blanks for the stressed test

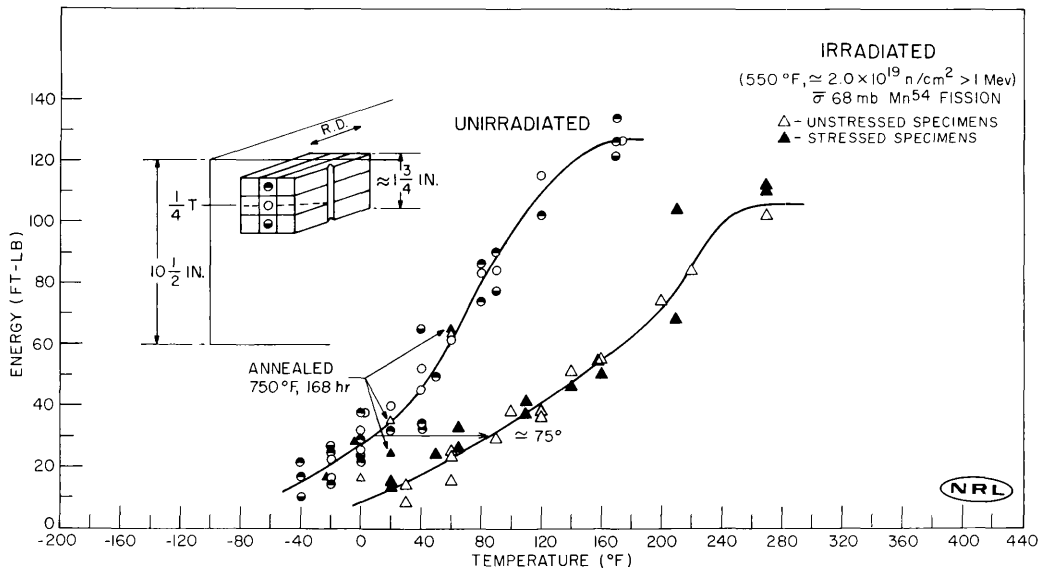

Fig. 10 - Comparison of the notch-ductility performance of 10-1/2-in. ASTM A302-B steel plate irradiated at 550°F in the stressed and unstressed conditions

bars were taken in two layers about the quarter thickness location as a material conservation measure. Reference specimens were also taken from this volume of metal to allow an exact comparison of radiation response. Following neutron exposure, specimens representing the unstressed condition denoted a transition temperature increase of 75°F (open triangles and curve). Significantly, the data for the stressed steel condition (closed triangles) fall closely about the transition-temperature curve for the unstressed condition, thus again demonstrating that an applied tensile stress during the irradiation neither enhances or reduces the radiation embrittlement sensitivity of pressure vessel steel.

The response of both steel conditions to postirradiation annealing at 750°F for 168 hours was also explored in this second study. The selection of the particular conditions for annealing was intended to reveal any changes in required conditions for, or the enhancement of, such elevated temperature metallurgical instabilities as temper embrittlement. As shown in Fig. 10, however, full transition temperature recovery was observed for both material conditions, thereby suggesting that stress during irradiation does not alter postirradiation annealing response.

The observation of a relatively small but detectable strain loss in the irradiation of specimens of the A302-B steel at 550°F as well as in the irradiation of the A350-LF1 (Modified) steel at 430°F may be taken as a general effect of radiation on stress relaxation characteristics of metals and alloys, as proposed by some investigators (10-16). However, an "apparent loss" in strain could also result from exceeding the elastic load limit of the test material at some point, either in preirradiation assembly or in preirradiation oven tests of thermal expansion conditions. A review of the conditions of the experiment involving A350-LF1 (Modified) steel indeed suggests this to be the case, since the specimens were loaded into the plastic strain range when at temperature. If some plasticity is introduced, this permanent set, though part of the total specimen strain, would not be accounted for in postirradiation strain measurements. Subsequently, a loss in recorded strain values in some cases may be fictitious.

Additional Investigations Planned

The A350-LF1 (Modified) steel forging and the A302-B steel plate materials discussed thus far represent respectively tempered ferrite and tempered upper bainite microstructures common to current reactor pressure vessel steel heat treatment practices. However, a new series of steels is being considered for advanced reactor designs in which higher strength levels will be provided by steel chemistry in combination with water (pressurized spray) quenching and tempering treatments in the interest of developing tempered martensite and/or lower bainite structures. These steels, particularly those of the Ni-Cr-Mo type, have a general susceptibility in their unirradiated condition to temper embrittlement at temperatures in excess of 700°F. However, the effects of radiation on the C-curve for temper-embrittlement formation has not been established. Similarly, the potential for other metallurgical instabilities has not been fully defined for the irradiated condition of these steels. Therefore, it is considered that parallel investigations of the response of the quenched and tempered steels to stress and radiation at temperatures in the ranges of 550° to 650°F and of 700° to 800°F will be necessary and are planned as part of the continuing NRL irradiation-effects program.

DISCUSSION AND SUMMARY OF OBSERVATIONS

The work by Nichols and Harries (5) and by Reynolds (6), together with recent experimental evaluations by NRL, constitute the sum of known investigations of the effects of static stress and radiation on the behavior of mechanical properties of carbon and low alloy steels. While representing significant progress in a highly questioned area, these limited studies have not fully answered the uncertainties of stress effects; therefore, the state of the art is still considered one of exploration and qualification. In particular, early attempts toward defining the significance of service stress in a radiation environment were shown to have points of deficiency either in terms of experimental approach or in equipment design. Later studies have sought to overcome such objections; however, several areas of additional, potentially fruitful research have been pointed out. In all cases, research assessments have relied upon Charpy-V notch ductility determinations as a highly sensitive indicator of any change in irradiation response.

Experimental results to date have not revealed either a significant effect of stress on subsequent irradiation response of carbon or low alloy steels, or a synergistic effect of stress and radiation on performance given by mechanical properties. NRL studies have additionally shown that stress during irradiation probably does not have a general effect on postirradiation annealing response.

Part 2

EFFECT OF NEUTRON RADIATION ON CYCLIC STRESS PERFORMANCE OF STRUCTURAL STEELS

PROBLEM DEFINITION AND ALTERNATE EXPERIMENTAL APPROACHES

Many engineering and regulatory groups have raised the question of potential radiation-induced alterations of the capacity of structural steels to withstand uniaxial and multi-axial stresses of a cyclic or fluctuating nature. In a recent report (17) from the design-oriented ASME Research Committee on the Effects of Radiation on Materials, for example, it was stated that "Low cycle fatigue (of irradiated pressure vessel steels) is of most immediate interest to the vessel designer," wherein low-cycle fatigue failure was defined as that which occurs in less than 100,000 load cycles. In contrast to low-cycle fatigue properties, the design requirement is less critical for high-cycle fatigue data in which stress rather than strain is the controlling variable. Experimental investigations of fatigue properties of irradiated steels and other alloys will be seen to reflect the Committee's opinion as to major needs and appropriate research emphasis.

Investigations undertaken to date have been largely exploratory in nature and are seeking to qualify both the direction of possible radiation-induced changes in low-cycle fatigue resistance and the general extent of such changes. From the design standpoint, the experimenters have set out to determine (a) if an engineering problem exists, and (b) if one exists, how serious the changes in fatigue properties are. Unfortunately, pursuance of this goal, as in studies of static stress effects, is difficult from the technical standpoint. Not only must the singular effects of irradiation be established, but also the combined or simultaneous effects of a nuclear and physical stress environment must be explored. Thus, the experimenter is faced with the ultimate need for actual in-reactor testing of specimens.

The primary question regarding the singular effects of irradiation stems from the increase in yield and tensile strength concurrent with a loss in tensile ductility. A "crossover" between preirradiation and irradiation test data curves representing strain versus cycles of failure could be projected as the outcome of this dual effect. Further, the point of crossover would delineate the change in competing mechanisms from that of ductility-controlled to strength-controlled fatigue resistance. As will be discussed, experimental results from various exploratory programs suggest that this crossover might, in fact, be the normal irradiation response.

One distinction which must be made prior to the presentation and discussion of research data involves the separate properties of fatigue crack initiation versus fatigue crack propagation rates. Most, if not all, radiation research efforts to be described may be classified as primarily studies of fatigue crack initiation. Parallel studies of fatigue crack propagation and growth rates, while recognized as essential to component design and operation, have not been reported upon in the open literature to the authors' knowledge. This area of future effort will be discussed further in the closing section.

EXPERIMENTAL RESULTS FROM PRE-POST IRRADIATION FATIGUE TEST PROGRAMS

Most research programs for assessing the fatigue-cycling response of steels when placed in a reactor service environment have taken the two-step experimental approach, or that of sample irradiation without stress followed by out-of-reactor testing. In addition to allowing volume testing, this approach has decided advantages in terms of experimental simplicity and background experience from nonradiation testing. Both full reverse bend tests and uniaxial tension-type tests have been performed at individual sites. For ease in presentation, the efforts and results of various laboratories are discussed separately.

Bettis Atomic Power Laboratory

Gibbons, Mikoleit, and O'Donnell have reported (18) on Bettis studies of the effects of radiation on the low-cycle fatigue properties of two pressure vessel steels, including a low alloy manganese-molybdenum A302-B steel and a carbon-silicon A212-B steel. Two heats of the A302-B composition and one heat of the A212-B steel were investigated in which smooth and notched specimens irradiated at 450° to 470°F to neutron fluences up to 1×10^{19} n/cm^2 >1 Mev were strain cycled at room temperature in axial fatigue (tension-compression). The test equipment used in these studies provided a reversed, constant deflection amplitude controlled by an eccentric system and allowed testing rates up to 900 cycles per hour.

The investigators found that for all of the materials tested using specimens having a smooth contour, "irradiation tended to raise the endurance limit and reduce the low-cycle fatigue life somewhat (18)." However, no appreciable difference was observed between the fatigue life of sharply notched (0.002-in. notch radius) versus less sharply notched (0.01-in. notch radius) specimens.

Reactor Materials Laboratory, Culcheth (UKAEA)

In a very recent report, Wood and Johnson (19) of the United Kingdom Atomic Energy Authority presented results from a detailed study of two steels tested in reverse bending at ambient temperature after irradiation at 140°C ± 10° to neutron fluence levels up to 1×10^{20} n/cm^2 >1 Mev. The objective of these tests was to examine the potential low-cycle fatigue performance of selected steels employed or considered for the internal liner of prestressed concrete pressure vessels. The two steels involved were a low carbon, high manganese steel having a 28 tons/in.^{2} UTS and a low alloy steel having a UTS of 36 tons/in.^{2}. The first steel was tested in two fluence conditions, 1×10^{19} and 1×10^{20} n/cm^2 >1 Mev, while the low alloy steel was tested in the higher fluence condition only. All postirradiation tests were performed at ambient temperature using 1/4 × 3/4 × 7 in. long specimens cycled at a rate of 3 cpm.

Figures 2 and 3 of Ref. 19 have been reproduced here as Figs. 11 and 12; the data clearly demonstrate the possibility suggested earlier of a "crossover" of curves representing preirradiation and postirradiation test data. The point of crossover, signifying the condition at which strength rather than ductility becomes the controlling factor in fatigue response, is noted to be somewhat different for the two steels (3 percent versus 4.5 percent total strain range for the low alloy steel). However, the factor more critical to reactor design and operation is that the irradiation effect on fatigue life is not very marked for either of the two fluence conditions and that the small changes observed are in a beneficial direction beyond 5000 cycles to failure (or up to 2 to 3 percent total strain range). Furthermore, the beneficial effects increase with increasing neutron fluence.

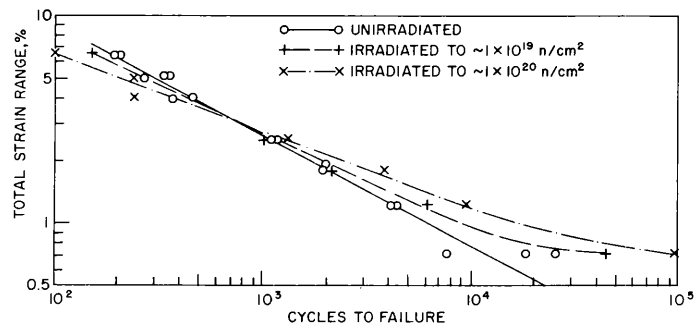

Fig. 11 - Effect of irradiation on the fatigue behavior of a low-C, high-Mn steel (19)

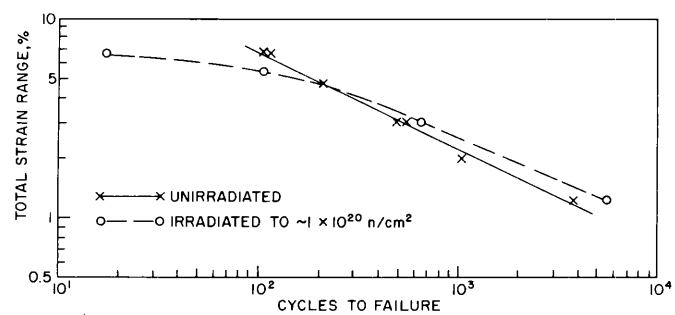

Fig. 12 - Effect of irradiation on the fatigue behavior of low-alloy steel (19)

The research by Wood and Johnson represents a significant contribution toward the understanding of fatigue performance of irradiated structural steels. The results of this investigation, however, do not fully qualify the potential effects of radiation on low-cycle fatigue properties, since the two-step approach was exercised. Additionally, it is seen that both the temperature of irradiation and the temperature of test are not all-encompassing experimental conditions, particularly when the service conditions of light-water-moderated power reactors are considered. From knowledge of notch ductility response during neutron exposure as well as of general fatigue behavior, the conditions of irradiation and test as applied for the specific study of liner steel performance are probably more severe than that which may be expected for water-cooled reactors operating in the range of 500° to 600°F. However, the investigator cannot be assured of this factor, or of the nonexistence of other contributing mechanisms until proven experimentally.

Idaho Nuclear Corporation Laboratories

Brinkman and Beeston (20) have examined the low-cycle fatigue properties of heavy section plate from single heats of A302-B and A212-B steel using 1/8-in.-gage diameter × 1/2-in.-gage length, smooth (unnotched) samples. The samples were tested in strain-range-controlled axial tension and compression at ambient temperature following irradiation at 60°C in contact with process water of the Materials Testing Reactor (MTR). In contrast to the studies of Wood and Johnson, the neutron fluence condition examined in this investigation was very low (approximately 4×10^{17} n/cm^2 >1 Mev). However, the results for A302-B steel in the test range of 10,000 to 100,000 cycles were interpreted by Brinkman

and Beeston as showing a slight reduction of fatigue life in the high strain range, together with a slight increase in fatigue life in the low strain range. The reduction in fatigue life in the region of 10,000 cycles to failure is not consistent with reported UKAEA observations for this number of cycles. For the A212-B steel, an effect of irradiation on fatigue behavior was not discernible within the test-data scatter.

OTHER STUDIES OF IRRADIATION-FATIGUE BEHAVIOR INVOLVING THE TWO-STEP APPROACH

Other known studies of low-cycle fatigue resistance of irradiated structural steels and alloys include investigations by R. D. Wylie and coworkers of the Southwest Research Institute (SwRI), undertaken as part of the Elk River Reactor (ERR) surveillance specimen program (21-23), and investigations by the Centre National d'Recherches Metallurgiques (CNRM), Liege, Belgium, by R. Salkin, in which the effects of fatigue cycling on the radiation embrittlement sensitivity of pressure vessel steels are being examined. An interim report of CNRM progress and observations has been issued (24).

The long-term SwRI program still underway involves heavy section A302-B, A105-Grade II, and A212-B steels, but only the A302-B and A105-Grade II steels have been included in the fatigue test series. Included in the objectives of the ERR vessel shell studies is an examination of the effects of prior fabrication history on the fatigue resistance of 3-in. A302-B plate. Postirradiation fatigue tests are by the uniaxial tension cycling method, and results have been obtained on some specimens irradiated in the ERR between December 1964 and April 1966. Although the test data were reported in advance of final neutron fluence determinations, it was found that the "exposure, however, has increased the fatigue strength of the A302-B specimens approximately 10 percent and that of the A105-Grade II specimens more than 20 percent (25)." This increase in life was shown for the low-cycle range (10^2 cycles) as well as for the high-cycle range (10^5 cycles). SwRI results from similar tests of irradiated dissimilar-weld fatigue specimens (weld overlayed A105-Grade II butt welded to Type 304 stainless steel) may be expected in the near future as well.

One other study now in progress is that of NRL, in which relative material response to pre-post irradiation testing versus dynamic in-reactor testing (during irradiation) will be determined and assessed. NRL plans are outlined later in this report.

EXPERIMENTAL RESULTS FROM DYNAMIC IN-REACTOR TESTING OF PRESSURE VESSEL STEELS

In the preceding section, it was shown that trends in the low-cycle fatigue properties behavior of irradiated reactor structural steels and alloys were becoming evident from the results of exploratory studies which had taken the indirect or two-step experimental approach. For full evaluation of the effects of radiation on fatigue properties, studies must necessarily couple radiation with simultaneous fatigue cycling. One special apparatus developed by NRL for the performance of dynamic in-reactor fatigue testing has been described previously (7). In brief, the apparatus provides for the simultaneous testing of 15 specimens in reverse bending at a selected temperature within the range of 300° to 700°F. The specimen is of the sheet-type, with a 1/16 × 3/16 in. cross section in the active test area; the criterion for fatigue life is the full failure and separation of the test specimen. Strain amplitude developed within each specimen is measured in the actual test apparatus following a minimum of 15 fully reversed cycles to establish hysteresis characteristics. Strain measurements in advance of radiation cycling as well as for control testing are made at room temperature, rather than at the elevated test temperature, to obtain the necessary reference points for data correlation. The results of the

initial experiment involving specimens from a 6-in. A302-B plate and 500°F test temperatures were presented as part of the formal description of the in-reactor test and control equipment (7).

Three additional in-reactor fatigue test evaluations have been performed (Figs. 13, 14, and 15) since the time of original publication (7). Figures 13 and 14 represent two assessments of one heat of A212-B steel (4-in. plate). The neutron fluence at failure ranged from 8.6 to 10.3 $\times 10^{18}$ n/cm^2 in the first study and from 10.3 to 11.8 $\times 10^{18}$ n/cm^2>1 Mev for specimens in the second study, which involved higher strain amplitudes. The separation of data points in these two figures suggests, for the failure range investigated, an improvement in the fatigue resistance of this particular plate occurring with irradiation testing. In this regard the more definite separation of data between irradiation and control tests in the second case may be related to the somewhat higher neutron fluence prior to specimen failures, or possibly to 'a threshold neutron fluence' for changes in fatigue resistance. In comparison with these results, similar in-reactor testing of two heats of A302-B steel at 500°F (7) and at 550°F (Fig. 15) did not reveal an effect of irradiation on fatigue resistance. The reason for the difference in fatigue resistance between the A212-B and A302-B steels is not immediately apparent; however, differences between the radiation embrittlement sensitivities of A212-B and A302-B steels have been found in independent studies involving 490°F exposures of Charpy-V specimens (26).

From the in-reactor test investigations, it may be tentatively concluded that dynamic fatigue cycling during irradiation does not produce effects grossly different from those observed with irradiation followed by postexposure cycling. However, the data from in-reactor testing obviously are limited and, for practical purposes, represent only two steel compositions among those of interest to the designer. As an alternative to performing costly and time-consuming in-reactor assessments on a routine basis, a program for specimen irradiation followed by postexposure cycling has been undertaken by NRL as a study supplemental to those dynamic test investigations already performed. The intent is to compare the low-cycle fatigue behavior of the two steels by both the direct and indirect experimental approaches. Material from the original test plate stock is being used for this correlation study. For the remote in-cell testing of irradiated specimens, a pneumatically driven unit was developed as shown, readied for tests, in Fig. 16. As with the in-reactor cycling assembly, the test method is one of reverse bending and is controlled by the magnitude of specimen deflection. Figure 17 is a close-up view of the specimen-actuator area, showing the placement of strain gages on the active test sections of the specimens. Prior to actual cycling, the gages are removed after making strain measurements and break detector wires are attached. In addition to the specimen design shown, a second design is being used which features a tapered test section by which strain at a given deflection can be made independent of the material yield strength. The current program, which involves both specimen types, includes tests at ambient temperature and at 500°F. Thus, an intercomparison with not only NRL prior tests, but also those of other laboratories, may be possible.

DISCUSSION AND SUMMARY OF OBSERVATIONS

The available experimental data, though very limited in view of the exploratory nature of tests to date, appear quite consistent in suggesting that fatigue resistance improves with irradiation for the strain range corresponding to N = 10^3 to 10^4 cycles or greater. At higher strains (N <10^3 cycles) the data would suggest that some reduction in fatigue life may occur with some steels as commensurate with the known loss in tensile ductility. The magnitude of this reduction in those steels so affected, however, would appear small or insignificant for neutron fluences currently projected for pressure vessels in service. The effect of irradiation on fatigue resistance, therefore, would not seem to be a problem with regard to reactor vessel design, at least for the steels which have been investigated.

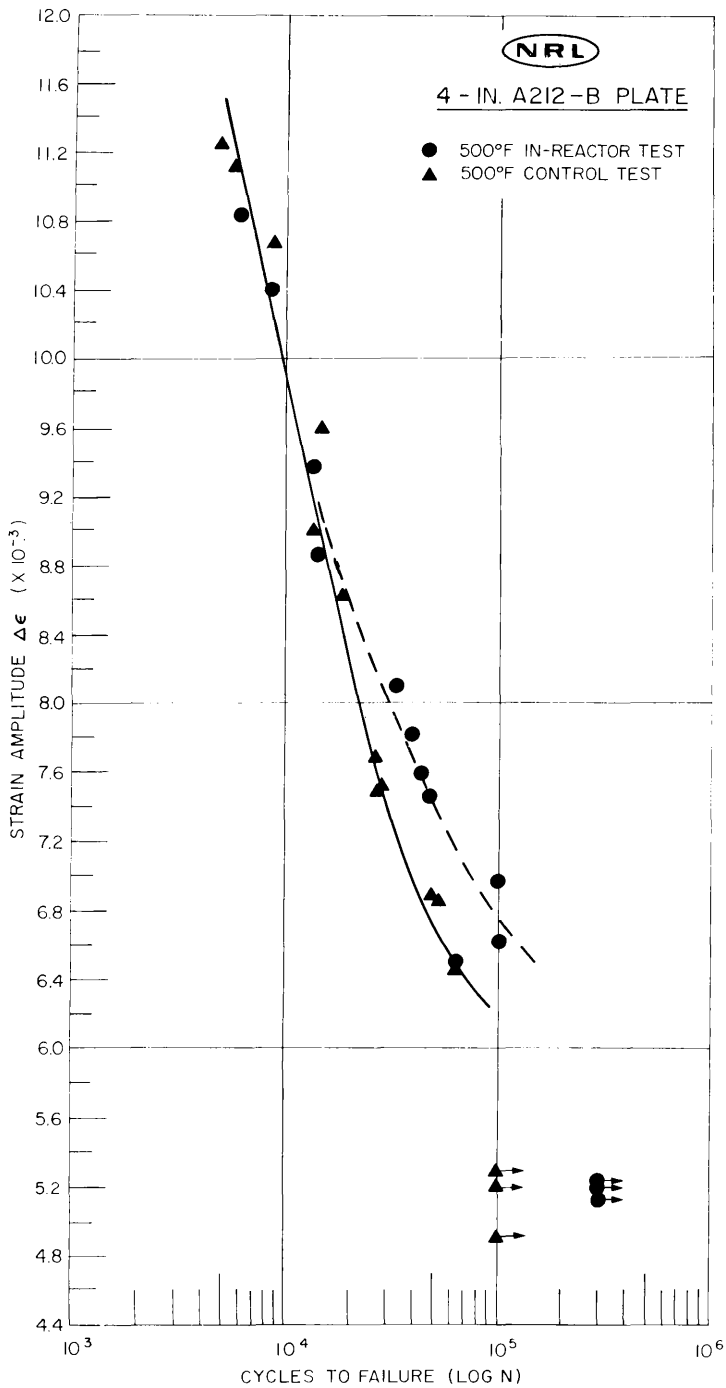

Fig. 13 - Comparison of fatigue data for 4-in. A212-B steel plate developed by in-reactor tests at 550°F, with the results of out-of-reactor control tests. The measurements of strain amplitude were at room temperature.

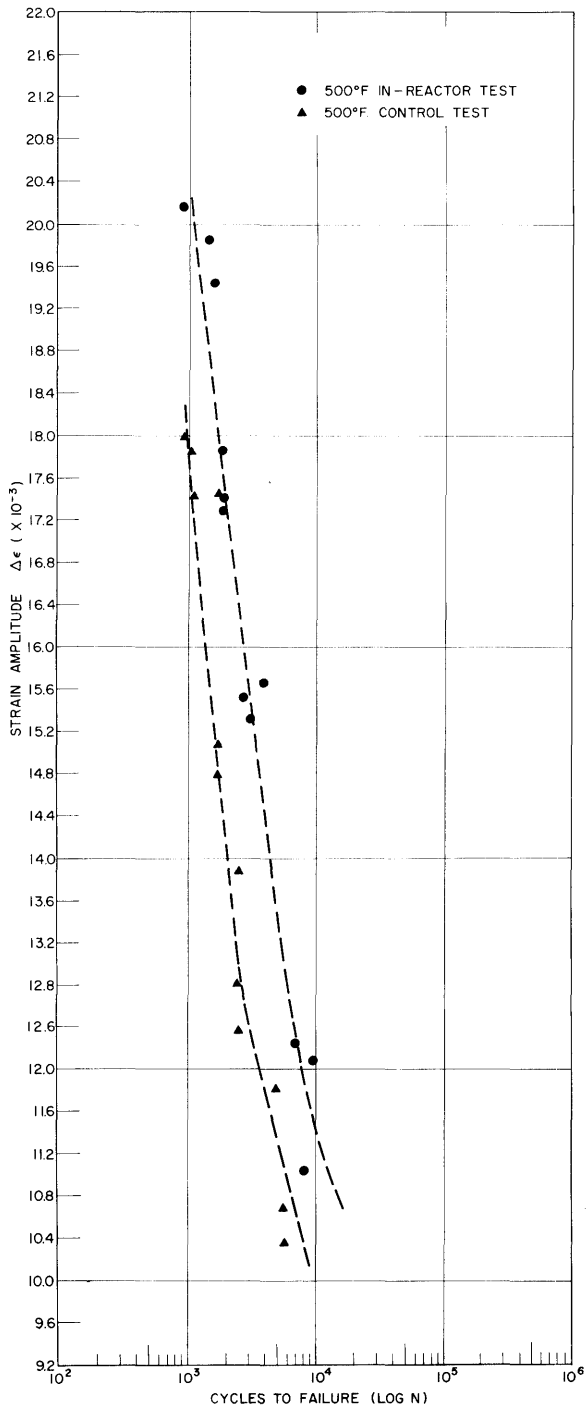

Fig. 14 - Comparison of fatigue data (at higher strain amplitudes) for 4-in. A212-B steel plate developed by in-reactor tests at 500°F, with the results of out-of-reactor control tests. Some suggestion of convergence at a higher strain amplitude can be noted.

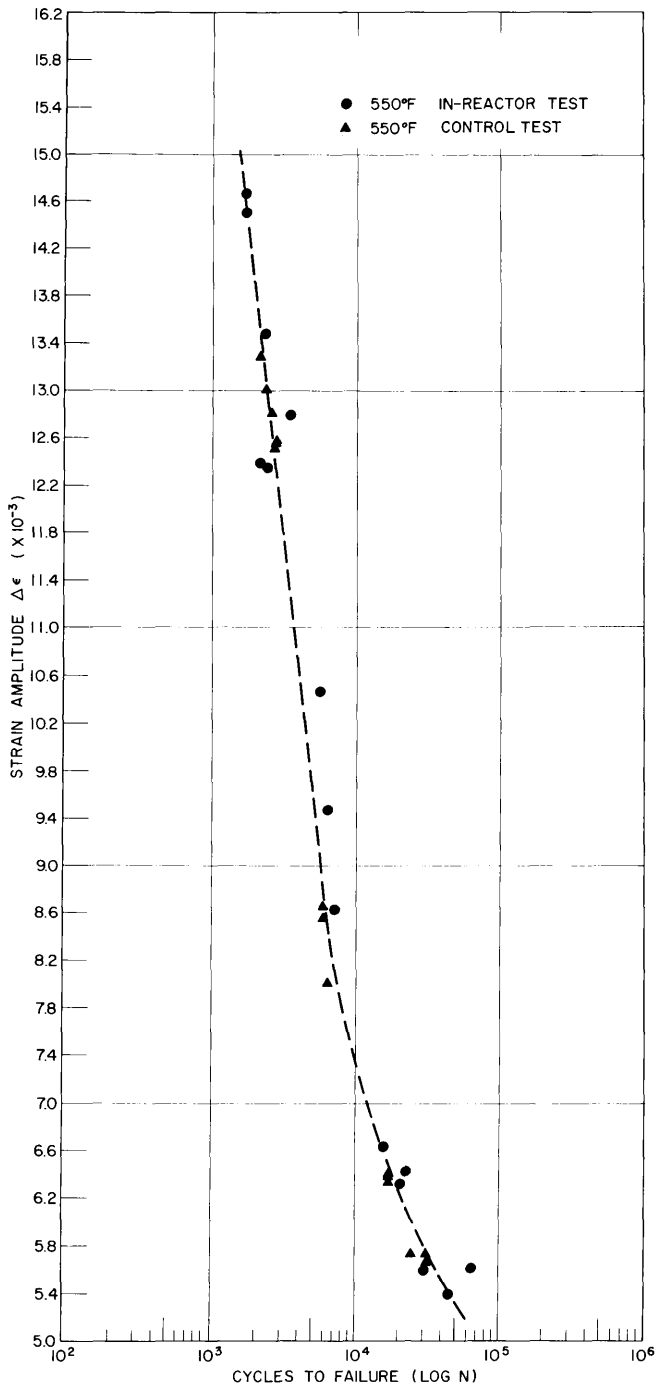

Fig. 15 - Comparison of fatigue data for 3-in. A302-B plate developed by in-reactor tests at 550°F with the results of out-of-reactor control tests. No evidence of an irradiation effect is apparent.

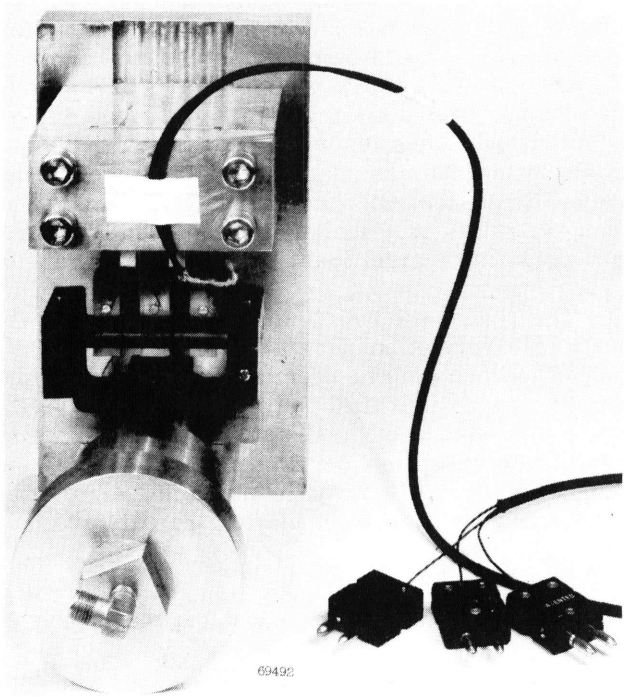

Fig. 16 - Hot-cell apparatus for remote testing of three reverse bend fatigue specimens simultaneously. For elevated temperature tests up to 600°F, the equipment is inserted into a specially adapted furnace. Split thermocouple wires shown spanning the specimen active test sections are used in combination with a potentiometer type recorder with thermocouple burnout features to monitor specimen failure (separation).

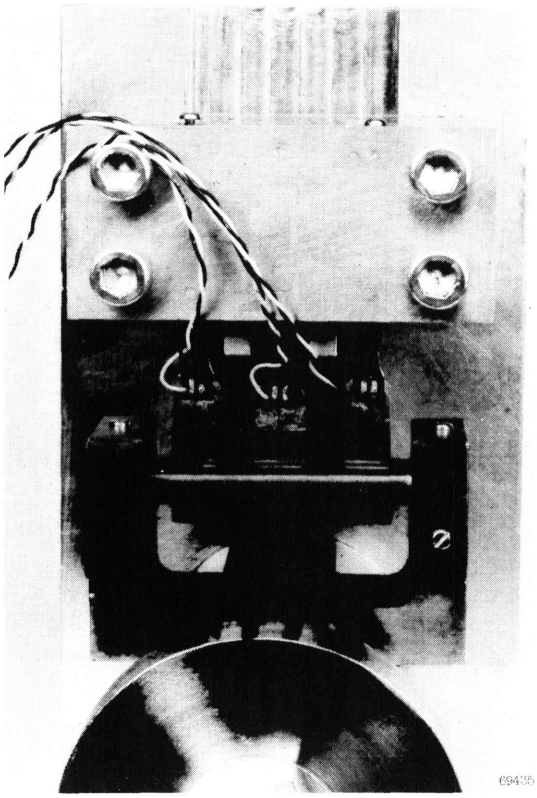

Fig. 17 - Hot-cell apparatus for remote testing of three reverse bend fatigue specimens simultaneously. A close-up view of the specimen actuator area shows the placement of etched foil strain gages on the active test sections. After making required measurements, the gages are stripped and the split thermocouple wires attached remotely.

The industry must recognize that the full potential for radiation-induced changes in fatigue resistance has not been realized, since experimental programs thus far have only studied crack-initiation characteristics. From a design standpoint, one usually considers a structure to contain cracks or flaws of a size equal to at least the lower threshold of the nondestructive testing means employed during fabrication or periodic in-service examination. Thus, experimental information on fatigue developed so far may not be the most critical assessment, or the most useful. However, crack-initiation aspects entailed one of the easiest approaches available to experimenters to determine if a greater fatigue problem exists in a nuclear environment than in a nonnuclear environment. Nonetheless, this work was a significant beginning in a long unexplored area.

The real issue which ultimately must be resolved is that of crack-propagation rates in nuclear versus nonnuclear environments. Only when this rate is known accurately can the fatigue problem be dismissed or design controlled. This knowledge will similarly permit establishment of realistic time intervals between periodic proof tests, as discussed in the following section. Hopefully, planning of studies of crack-propagation rates under irradiation will allow correlation with similar studies in the unirradiated state. In particular, influences of corrosion and aging embrittlement as found by several different laboratories may influence irradiation results and must be acknowledged in such investigations.

Part 3

EFFECTS OF PRESTRESS AND PROOF-TEST APPLICATIONS: CONVENTIONAL VESSEL OPERATIONS CONCERNING NUCLEAR AND NONNUCLEAR STRUCTURES

BACKGROUND

One of the primary requirements of nuclear pressure vessel operation is the prevention of unexpected catastrophic failure. To this end vessels are subjected to exacting design and fabrication procedures, including an exhaustive series of nondestructive tests. Still, after the vessel has been constructed, it is required to undergo a proof test. This test may consist of a hydrostatic pressurization to 1.25 times the design pressure, adjusted for any variation in allowable primary stress intensity due to differences between test and operating temperatures, as required by Section III of the ASME Boiler and Pressure Vessel Code. The hydrostatic test is necessary to provide a positive demonstration that the vessel will withstand operating pressure. In the event that some factor has been overlooked, the test places the possibility of a failure at a time and place where it can be most easily dealt with.

The proof test carries with it the implication that the structure will be free from failure possibilities until such time that the material properties have been degraded by aggressive environment, existing flaws have grown by fatigue, or the metal has become embrittled by radiation. It has, therefore, been argued that a vessel should be periodically proof tested throughout its service lifetime to demonstrate that environmental factors have not affected the vessel properties to the extent that failure is imminent. The counter argument is that periodic testing may damage the vessel by means of local plastic strain embrittlement or aging embrittlement in the region of a flaw. Also, a static proof test does not guard against a dynamic fracture initiation caused by shock loading during operation or by a sudden jump in crack-growth rate at the tip of a flaw embrittled by environmental effects. These and other factors concerning proof testing are examined in the following sections. The ensuing discussion is directed at vessels that have been thermally stress relieved.

REASONS FOR PROOF TESTING

At this time the potential advantages of the initial proof test above the nil-ductility transition (NDT) temperature appear to outweigh the potential disadvantages, and this test should always be performed on a nuclear vessel. Its successful completion provides visible evidence that the vessel will withstand operating pressure. In addition, it is likely that such pressurization causes a favorable redistribution of stress at local discontinuities and around flaws where localized yielding takes place during the test. Upon depressurization, residual compressive stresses exist at such points which must first be overcome upon subsequent repressurization, before further yielding can occur. Since this repressurization during vessel operation normally takes place at a lower level, there should be no reyielding. The proof test is prima facie evidence that critical flaws do not exist at the test temperature and provides the basis for extended operation at lower pressure before any existing flaws that were just subcritical at the test pressure would grow to critical size by fatigue at the lower operating stress conditions. This reasoning assumes a knowledge of the flaw growth rate.

Nuclear vessels built today are designed for operation in the temperature region above NDT. However, one of the major reasons for considering periodic proof testing comes into play when the initial NDT is raised by irradiation to such an extent that either vessel startup and shutdown procedures requiring pressurized operations must be allowed to proceed near and below the NDT or the vessel must be removed from service. To cope with this situation, it is sometimes stipulated that a prior proof test above NDT has blunted cracks to an extent that brittle failure will not occur at any temperature below NDT until the proof-test stress values are exceeded, provided, of course, that the operating and proof-stress patterns are identical. This procedure is known as warm prestressing.

For the majority of nuclear vessels that are always operated above their NDT temperatures, there are arguments both for and against further periodic proof testing above NDT. One of the major reasons for such a procedure is the uncertainty that exists in nondestructive testing means. There is always the possibility that a large flaw just subcritical at the proof-test pressure may have been overlooked. Presumably a failure resulting from this flaw would occur during a subsequent proof test, after the flaw had grown by fatigue. However, once complete reliance can be placed in nondestructive tests disclosing all flaws over, say, a few inches for the largest pressurized water reactor (PWR) vessels, then there seems to be little chance for flaws to grow undetected to critical sizes of the order of one to two times the wall thickness by fatigue. But until such time, the periodic test is a good means to ferret out potential failure situations above NDT.

Crooker (27) has examined flaw growth rates for A302-B in a simulated boiling-water environment using a Lehigh-type bend test. Specimens were cycled in reverse bending to strain ranges typical of those measured in the region of flaws found in model tests of vessels sponsored by the Pressure Vessel Research Committee (PVRC) (28). The largest flaw growth rate found for cracks up to 2 in. in length, using the bend specimens, was about 10^{-3} in./cycle at a total strain range of 6000 μin./in. This range takes into account a factor for the effect of mean strain. If the above conditions are typical of those in a vessel nozzle region, then there is not much reason to believe that fatigue alone will cause appreciable crack growth over several hundred full power cycles during the vessel lifetime. Of course, as mentioned earlier, the synergistic effects of irradiation and possible corrosion on crack-growth rates must be further examined. It is also realized that a size effect associated with larger flaws and section sizes may result in higher flaw-growth rates than that implied above.

For vessels operated and proof tested above NDT, it is possible that such periodic testing above operating or design limits may cause some additional fatigue growth of flaws, but even this effect has never been shown. Indeed, it has been argued that proof testing can actually inhibit flaw growth by providing compressive residual stresses. Besides fatigue, another source of material degradation is embrittlement at the flaw tip (discussed later). However, little chance for brittle fracture exists above the NDT, and there seems to be no mechanism by which locally embrittled regions about the crack boundary can offset the inherent dynamic propagation fracture toughness of metal in the ductile regime. Here the most probable source of failure is leakage from a through-thickness flaw. Complete rupture could nevertheless occur by plastic instability resulting, for example, from bulging around an extremely large flaw. A flaw of such length (two to three times the wall thickness) is most improbable in a vessel which was thoroughly inspected before service.

REASONS AGAINST PROOF TESTING

The most interesting operation conditions from a fracture viewpoint arise when (a) the proof test is performed above the NDT and vessel operation is anticipated below NDT (warm prestressing), or when (b) both vessel operation and proof testing take place below NDT. There

are presently advocates of both procedures for nuclear vessels. For irradiated vessels, Holt (29) has suggested a proof test in the brittle as well as the ductile range. Also, pressurized operations for some vessels have been forced below the NDT by irradiation, so the following discussion is not removed from reality. The alarming fact concerning vessel operation below NDT is the risk of failure at elastic stress levels with small flaws of the kind that may escape nondestructive tests on the largest PWR vessels. Two potential sources of trouble will be singled out for a vessel that is operated below the NDT and has been proof tested as a means of guaranteeing integrity in this region. These are: (a) strain-rate effects, and (b) aging embrittlement. A subsequent section discussing the use of warm prestressing for operations below NDT is also relevant.

Strain Rate Effects

The nuclear pressure vessel materials (notably A302-B steel) are known to be strain-rate sensitive, meaning that they will exhibit a lower toughness when strained rapidly as opposed to static toughness. Since a proof-test loading is conducted in a "static" manner, it obviously cannot guarantee fracture safety under dynamic conditions for operation below NDT. However, this lower dynamic toughness would not seem to be a consideration, since vessel operating stresses are normally applied in a static manner. Nevertheless, two situations could trigger a failure in a vessel that has successfully passed the hydrostatic test. First, any type of shock loading, such as water hammer, may trigger an existing flaw. Secondly, a flaw whose tip has been locally embrittled could result in a sudden jump of crack growth through the embrittled region, causing the unaffected material to revert to its lower dynamic toughness. In other words, the properties of a small amount of metal bordering the crack can control the fracture below NDT.

First, consider the strain-rate effects related directly to dynamic loading. Shoemaker (30) has studied the embrittlement of A201-B steel caused by small-amplitude stress waves superimposed on a static loading condition. Tests were conducted with 3/4-in.-thick edge notch and bend specimens in the region below NDT where linear elastic fracture mechanics applies. The data show an average of 26-percent reduction in toughness, expressed in terms of K_{Ic}. This "embrittlement" was caused by the transient load due to the small stress pulse, in combination with the resultant strain-rate increase. However, the transient strain elevation accompanying the reduction in toughness was only 40 μin./in.; therefore this reduction is attributed to the dynamic effects of the stress pulse rather than to the elevation of the local stress field. Similarly, Nibbering (31) has demonstrated that the fracture strength of beams of ship steel may be lowered at least 30 percent by the superposition of a pulse loading on a static load. Beams were statically loaded to a stress level 30 percent less than that required to cause fracture using slow loading. A small weight dropped on the beam was then sufficient to cause fracture.

The second way of achieving a dynamic situation is through local embrittlement. Such embrittlement at the notch tip may trigger a failure at pressures below the proof stress level for below-NDT operation. One such form of embrittlement is aging embrittlement, which will be discussed in the following section. This form of embrittlement is considered by some to have been a mechanism leading to a recent British boiler drum failure (32). The Ducol W.30 vessel failed in a brittle manner while undergoing a preservice hydrotest below NDT. The failure pressure was somewhat less than that withstood under several previous hydrotests. The failure initiated from a large crack propagated during the stress relief operation. Possibly, the crack tip was plastically strained during previous hydrotests and aged during the time interval between. The process may have embrittled the flaw tip to provide a rapid burst of crack growth and thus trigger the failure.

It is not unreasonable to question why one would be concerned with the dynamics of crack initiation when vessels (and thus the flaws) are stressed very slowly in normal operation. In answer to this, Krafft (33) implies that contrary to normal expectations a

very slow forward crack velocity is in fact equivalent to an enormous rate of loading of a stationary crack, and that the distance such growth needs to occur to trigger fracture at dynamic toughness levels is quite minute. He envisions the material subdividing into ductile cells of diameter d_T (10 to 100 μin.) by advance nucleation around inclusions. These cells, or "process zones," are considered ligaments which must be ruptured to permit advance of the crack front to the next ligament.

Reference 33 gives the relationship between strain rate $\dot{\epsilon}$ at the process zone and crack front velocity V_c in terms of applied plane-strain stress intensity factor K_I, process zone size d_T, and modulus E as

$$\dot{\epsilon} = \frac{\dot{K}_I}{\sqrt{2\pi}\, E\, d_T^{\,1/2}} + \frac{V_c K_I}{\sqrt{2\pi}\, E\, d_T^{\,3/2}}. \tag{1}$$

From Eq. (1) it can be shown that the following relationship exists between the crack front velocity and loading rate of the overall structure $\dot{K}$ (in fracture-mechanics terms).

$$V_c = \frac{2d_T}{K_I}\dot{K}_I. \tag{2}$$

The $\dot{K}$ typical of a Charpy-V test is 10^8 psi $\sqrt{\text{in.}}$/sec. Using this value along with a d_T of 50 μin. for mild steel and a K_I of 50×10^4 psi $\sqrt{\text{in.}}$ typical of A302-B steel below NDT, Eq. (2) shows that a crack speed of 0.2 in./sec is equivalent to strain rates experienced in a Charpy-V test. It is this strain at the crack front which controls the fracture, and not the nominal strain rate of the complete vessel. The described crack velocity could be attained by a sudden burst of crack speed through a locally embrittled region. These conditions could, for example, be a jump of 0.001 in. in crack length in a millisecond, which is not unreasonable. The distance which a crack would have to jump is presently unspecified, but it may be as small as d_T. The fact that such a burst of crack speed would be almost impossible to preclude in a large structure must dictate a fracture control philosophy (including proof-testing) based on dynamic levels of fracture toughness.

In the region below NDT, data confirming the decrease in dynamic toughness are few, since this area has been relatively unexplored. However, Eftis and Krafft (34) have shown a general decrease in crack-initiation toughness with increased rate of loading. In fracture mechanics terms, they found a 50-percent decrease in the dynamic plane-strain fracture toughness, K_{Ic}, over the "statically" determined values using slow loading. The implications of such a drastic decrease in toughness become apparent if operation below NDT is contemplated. Consider typical vessels such as those in the Yankee or Indianpoint Power Stations, whose working pressures are about 55 percent of the hydrostatic test pressures. Here a 50-percent decrease in toughness due to dynamic effects could mean failure from a crack that was just subcritical above NDT, if full-pressure operation was attempted below NDT. Above NDT the material is tough enough to absorb any small burst at the crack front due to a locally embrittled region, and the argument of dynamicism would not apply.

Aging Embrittlement

In the preceding section, it was shown that a small burst of crack growth from the tip of a flaw in a locally embrittled region could cause a vessel to display a lower toughness below NDT. One such possible source of embrittlement is aging embrittlement, such as may have resulted in the previously described vessel failure. To explore this phenomenon, the authors have conducted a preliminary experiment using A302-B steel subjected to conditions believed conducive to aging embrittlement.

Material – Fracture mechanics specimens were cut from a 4-in. plate of A302-B steel purchased for use in the AEC coordinated program on the radiation effects on reactor structural materials. The plate was given a conventional heat treatment for reactor pressure vessel applications, but was not from a vacuum degassed heat. Since the 4-in. section is not as thick as most PWR and BWR vessels, the properties cannot be expected to duplicate completely those of actual vessels or of the original 6-in. ASTM "reference steel" donated to the ASTM by the United States Steel Corporation. The NDT temperature of the plate was found to be -90°F from drop-weight specimens cut from the 1/4T and 3/4T locations. However, Charpy-V curves from material on both sides of the 1/4T and 3/4T planes indicate a gradient in properties through the thickness. From these data one may project an NDT for center material at least 50°F higher than that experimentally determined at the 1/4T and 3/4T locations. A linear variation in yield stress of about 10 percent was noted from one surface of the plate to the other. The 2-percent offset yield stress values were determined from 1/4-in.-diameter × 7/16-in.-long compression plugs loaded slowly in testing. Each plug was oriented with its axis perpendicular to the specimen fracture direction Yield stress values as a function of temperature are listed in Table 3.

Table 3
Yield Stress Values for A302-B Plate

Plate Location	Yield Stress (ksi)	Temperature (°F)
1/4T*	74.6	+75 (Room Temperature)
Center	75.5	
3/4T	81.0	
1/4T	83.9	-90
Center	86.5	
3/4T	92.5	
1/4T	94.4	-150
Center	95.0	
3/4T	98.2	
1/4T	122.7	-250
Center	118.0	
3/4T	124.9	

*T is plate thickness (4 in.)

Because of the observed variation in properties, three groups of specimens (type and configuration to be described) were taken from 1/4T, 3/4T, and center-plate locations. The specimen thickness dimension was parallel to the 4-in. plate thickness direction in all cases.

Specimen Design and Preparation – In order to assess the magnitude of any aging effect, it was decided to use a fracture mechanics test. Since such a test is sensitive to conditions at the flaw tip, it should disclose any drop in crack-initiation toughness because of local embrittlement at the notch tip. The specimen configuration employed is shown in Fig. 18. The dimensions (1 in. thickness T, 16 in. span S, 5 in. width W) chosen are the same as for the 1-in. Drop Weight Tear Test (DWTT) specimen used at NRL. This choice

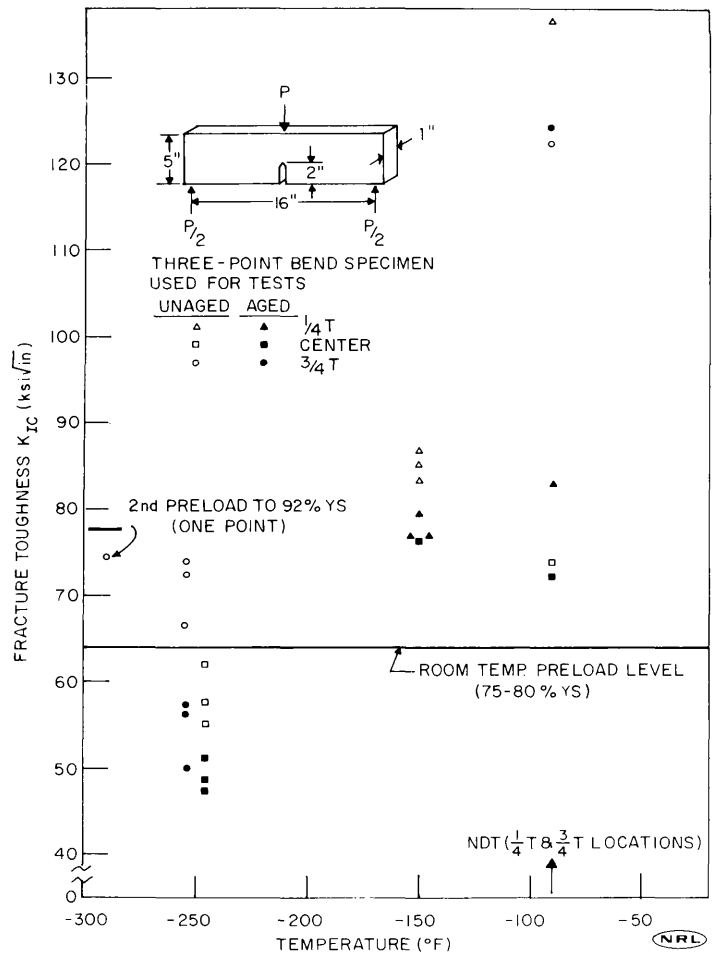

Fig. 18 - Fracture toughness of aged and unaged A302-B steel specimens at various temperatures

was made as a matter of convenience, in case it was later decided to run DWTT in addition to fracture mechanics tests. The specimen proportions are in general agreement with those recommended by the ASTM Committee E-24 on Fracture Testing of Metals.*

The three-point notched-bend specimens were subsequently cut with the notch parallel to the primary rolling direction. The notch consisted of a 1/16-in.-through-thickness slot sharpened at the bottom with a 60-degree mill cutter. The notches were terminated by fatigue cracks of length 0.125 to 0.150 in., making an overall crack dimension (a) of 2 in. The specimens were fatigued at room temperature (RT) as cantilever beams using a tension-tension cycle. This method imposed nominal bending stresses at the root of the notch that went from 13 percent to 30 percent of yield (about 8 to 19 ksi $\sqrt{\text{in.}}$ applied K level). Approximately 250,000 to 300,000 cycles were required to grow each fatigue crack.

*A recommended practice for plane-strain fracture toughness testing using a crack notch-bend specimen is in the course of preparation.

All specimens were then slowly preloaded in three-point bend to a nominal stress at the base of the notch of 75 to 80 percent YS (about 64 ksi $\sqrt{\text{in.}}$) and held for five minutes. One-half of the group was then aged in an oven at 550°F in air for 500 hours. The other half was left at RT. This procedure is representative of what actually happens in a nuclear vessel.

Test Procedures – All the specimens were loaded in a bend jig at a 4000 lb/min (11 ksi $\sqrt{\text{in.}}$/min) loading rate. A fracture mechanics type beam gage was inserted into the notch, and its output, along with the load output from the testing machine, was fed into an x-y recorder. Different test temperatures from RT to -290°F were obtained with liquid nitrogen applied through four chill blocks clamped on both sides of the specimen straddling the fracture path. The temperature was monitored by means of a thermocouple spot welded to the specimen. (Bench tests had initially shown essentially zero temperature variation in the test zone.)

Results and Discussion – The results of the tests at different temperatures are summarized in Fig. 18 in terms of the plane-strain stress intensity, K_{Ic}. This quantity was computed from the following formula of Srawley and Brown (35) for three-point bend.

$$K_{Ic} = \frac{3}{2} \frac{PS\sqrt{a}\,Y}{BW^2} \tag{3}$$

where

Y = function of (a/w) as given in Ref. 35

a = crack depth

B = thickness

W = width

S = span between supports

P = load

Since all the specimen dimensions (including the crack depth) were practically identical, the K_{Ic} value was essentially proportioned to the load at fracture instability. All of the load-deflection records with the exception of the upper three points at -90°F were linear, and the instability load is therefore also the maximum value. The foregoing three points do not represent valid fracture mechanics tests because of yielding, but are plotted for information using the maximum load in Eq. (3). Thus, essentially the same pattern of test points would be obtained in a plot of maximum load versus temperature.

Some comments are in order concerning the validity of the results as K_{Ic} values. The fact that all of the load-deflection records were linear up to fracture (with the exception of the three points noted) and that the nominal stress at the root of the notch at fracture was less than 0.9 yield stress are not sufficient criteria for valid results. In addition, the latest recommended practice from ASTM Committee E-24 requires that both a and B be less than 2.5 $(K_{Ic}/\sigma_{YS})^2$. These conditions are satisfied by all points at -250°F for both aged and unaged specimens. All of the remaining points satisfy the condition on a but not on B. Wessel (36) has suggested that a value of 1.5 $(K_{Ic}/\sigma_{YS})^2$ for B may be sufficient for A302-B. If this value is used, the aged points at -150°F are valid, while the unaged ones are less than 20 percent smaller than the thickness requirement. Also the single 1/4T material aged point at -90°F is 30 percent less than the thickness requirement.

As seen in Fig. 18, there is a trend for the unaged specimens to exhibit a higher fracture toughness than the aged ones for each plate location at temperatures far below NDT. Near the NDT temperature, the effect appears to diminish. At -250°F the average

K_{Ic} of three aged specimens from the 3/4T location shows a K_{Ic} 23 percent below the average of three unaged specimens. For tests from the center plate location at the same temperature, the aged specimens exhibit an average K_{Ic} 15 percent below the average unaged value. At -150°F the aged 1/4T results are 8 percent below the average unaged fracture toughness. Considerable scatter is noted at NDT, and no conclusions can be drawn. The upper three points at -90°F are not valid tests and should fall lower if plane strain conditions had been obtained.

Another point to be noted from the figure concerns the relationship of the fractures to the preload level. Some failures for both aged and unaged specimens fell below the preload level at -250°F, while at -150°F and higher they all fell above this level. To check this behavior at -250°F one unaged specimen from the 3/4T location that exhibited failures above the preload level at -250°F was preloaded again, but at this time to 92 percent of RT yield stress. At -290°F this specimen failed below the preload level. The reason for this behavior is unknown, and the results show a pattern to be expected for specimens without preload, namely, a rise in toughness with temperature. However, it is not unexpected that failures would occur at levels higher than the preload level, even at temperatures below NDT. This would be the case if the applied preload was less than the load required for failure below NDT in a virgin specimen. Evidently, this was the case, but no unpreloaded specimens were available to provide a check.

These preliminary results point out that there appears to be an effect of aging embrittlement on this plate of A302-B steel which should be further investigated. Additional tests are required before general conclusions can be drawn. However, one implication concerning vessel operation below NDT can be made. If a vessel is proof tested below NDT to 1.25 times its design pressure and there exists a flaw that is just subcritical, then the scatterband of data indicates that the aging-embrittlement effect could cause failure upon subsequent pressurization at the same temperature at 85 percent of design pressure. This pressure approaches the operating level of a nuclear vessel. The authors plan to continue this investigation using another heat of A302-B and other steels. One point that will be investigated is the possibility that this form of embrittlement will raise the NDT temperature.

WARM PRESTRESSING

Vessel operation below NDT can be hazardous, and for this reason warm prestressing above NDT is being considered as a means of guarding against fracture for this type of operation. Understandably, the motivation for a proof test at or below operating temperature (i.e., below NDT) is also strong. Therefore, before discussing warm prestressing, consider first what the implications of a hydrotest below NDT are. The purpose of such a test is to disclose any flaws that are just subcritical at operating conditions. As described in the following section, the chances of having critical flaws at the test pressure are real.

Proof Test Below NDT

To estimate a critical crack size below NDT, one may use a fracture mechanics approach if plane strain conditions prevail. The depth of a critical elliptic surface flaw may be computed from the following equations of Irwin found in Ref. 37.

$$a = \frac{Q}{1.21\pi}\left(\frac{K_{Ic}}{\sigma}\right)^2 \tag{4}$$

where

a = crack depth

Q = flaw shape parameter = $\phi^2 - 0.212\,(\sigma/\sigma_{YS})^2$

ϕ = complete elliptical integral of the 2nd kind

σ = gross stress around flaw

σ_{YS} = yield stress

K_{Ic} = plane-strain stress intensity factor

For a flaw residing in the cylindrical portions of a vessel, where a uniform hoop stress prevails, a critical crack size can be calculated assuming a hoop stress of two-thirds yield stress around the flaw. Such a stress level is reasonable when out of roundness and possibly residual stresses are considered in addition to the nominal hoop stress. A reasonable value for K_{Ic} below NDT is 40 ksi $\sqrt{\text{in.}}$, from Wessel (36). Assuming a flaw length-to-depth ratio of one half and a yield stress of 60 ksi for A302-B steel, Eq. (4) indicates a critical flaw 0.58 in. deep by 1.16 in. long.

Next, consider a flaw residing in the nozzle region. This area is more likely for fracture initiation than the uniform shell region. The inherent nozzle geometry results in stress concentrations and stress gradients. This situation is further aggravated by bending stresses due to pipe reaction loads on the nozzle. This set of conditions makes it more difficult to determine a gross stress surrounding the flaw as required to apply Eq. (4) correctly. The maximum nozzle stress magnitudes are easily at the yield stress level. If one assumes that this yielded region exists for a small but sufficient distance around a flaw, then Eq. (4) should be a good approximation for the critical flaw size, Using the above values for K_{Ic} and yield stress, and assuming a yield stress loading. Eq. (4) gives a critical crack of only 1/4 in. deep by 1/2 in. long. If the toughness were greater, or the applied stress less, the critical crack size would go up as $(K_{Ic}/\sigma_{YS})^2$. For example, if K_{Ic} were really 80 ksi $\sqrt{\text{in.}}$ with σ_{YS} = 60 ksi, then the critical nozzle flaw would be 1 in. deep by 2 in. long. These critical flaw sizes are of a magnitude that may escape nondestructive tests for a large (12-in.-thick) PWR vessel. Further, if the proof-test results in aging embrittlement at the tip of the flaw, failure is even more likely. Finally, the hydrostatic test below NDT cannot guarantee safety against dynamic crack initiation, as discussed previously; therefore, the lower dynamic K_{Ic} must be considered in flaw size calculations. Unless a material of higher fracture toughness than 40 ksi $\sqrt{\text{in.}}$ below NDT is used, a proof test at this temperature is not considered advisable.

Proof Test Above NDT - Warm Prestress

As a means of circumventing the undesirable aspects of operation below NDT, it has been suggested that the vessel first be hydrostatically tested above NDT to insure safe operation at the lower temperature. The actual effect of this operation on the material around the crack flaw has not been completely analyzed. One possible result is a blunting of the notch tip. There may be no physical blunting, but plastic deformation can make the effective radius of the crack tip larger. Also, compressive residual stresses remain after depressurization, and these must first be overcome on subsequent pressurization to initiate fracture. In any event, it has been known for some time that warm prestressing can carry with it a benefit concerning pressurization at lower temperatures. Some experimental work leading to this conclusion is discussed in the following paragraphs.

Brothers and Yukawa (38) have investigated the effect of warm prestressing at various levels on the notch-bend fracture properties of a Ni-Mo-V steel. The test bars were fatigue cracked and preloaded in three-point bend to various levels of nominal stress at the

root of the notch, ranging from below yield to above ultimate stress. The bars were preloaded with the notch in tension at a temperature of 90°F above the Charpy 50-percent fracture appearance transition temperature (FATT) and then tested from 200° to 250°F below FATT in slow bend. The results show a general increase in the fracture load with the increase in prestress level and indicate as much as 150 percent increase in fracture load for prestressed bars over those not receiving prestress. This trend continues up to some optimum level of prestress which is in excess of ultimate stress at the base of the notch. Preloads above the optimum levels were accompanied by cracking during the prestress cycle and resulted in lower fracture loads than the original preload value. These investigators emphasized the localized nature of the warm prestressing effect around the flaw region by investigating fracture appearance, load-deflection records, and energy requirements for crack propagation. They also indicated that some plane carbon steels susceptible to strain aging may be adversely affected by the warm prestressing procedure.

Kihara and Masubuchi (39) have conducted similar tests in which the effect of warm preloading on plates of mild steel having a weld joint with a sharp flaw was examined. The plates were preloaded above the crack arrest temperature to various stress levels below yield and later tested below this temperature. Results showed an increase in the value of fracture stress at low temperatures, with no failures occurring below the warm preload level.

Bevitt, et al. (40), burst large steel pressure vessels of 1-in. wall thickness. One particular 5-ft.-diameter vessel was pressurized above NDT to a hoop stress of yield level. This test resulted in a 3/4-in. crack extension from the original 6-in.-long through flaw. Later the vessel was cooled below NDT and pressurized to the same level without failure. On the other hand, a similar vessel that had not received the warm prestress failed at 84 percent of the pressure withstood by the previous vessel at the same temperature below NDT. From this result, Bevitt and coworkers concluded that warm prestress is a reasonable method to assure complete safety in the period between proof tests, provided the stress patterns are the same in operation as in the proof test and that the test is carried out at a high enough pressure to allow a margin for the expected crack growth between proof tests.

The preceding evidence shows that the warm prestressing effect does exist, under certain conditions, but the real question is how to make use of it in a nuclear vessel operated below NDT. Clearly, there are areas that must be carefully examined before this technique can be relied upon. Some of these areas are described in the following paragraphs, and they indicate that one must proceed with utmost caution when attempting to exploit the benefits of warm prestress.

1. The actual effect of the proof stress on the material around the notch tip has not been fully defined. If the effect is localized, as suggested by Brothers and Yukawa (38), then one must explore the result of having a fatigue crack grow (between proof tests) to the outside of the area of benefit, thereby resulting in the same type of vessel fracture behavior as would occur without warm prestress. Along this line, a great deal of work remains to be done in determining fatigue-crack growth rates, especially in a nuclear environment. Embrittlement at the crack tip caused by aging or other mechanisms may increase the crack growth rate.

2. The effect of environment on crack tip behavior may cause local embrittlement, such as aging embrittlement. If this behavior then provides a source of dynamic crack initiation, the static warm proof test does not provide positive assurance against failure.

3. As noted by Brothers and Yukawa, if the strain around the flaw is not known or becomes too large, localized cracking may result. This, in turn, reduces the expected benefit of warm prestressing.

4. In a discussion following Ref. 38, Sankey notes that the benefits of warm prestress will be reduced when plane strain conditions prevail. For large rotors not at a condition of complete plane strain, he indicates a 25-percent drop in the notch strength over the results obtained in Ref. (38). Conditions of complete plane strain would seem to indicate still lower values.

5. A different stress distribution during vessel operation from that during preload must also be considered. Items such as pipe reaction loads at nozzles and thermal stresses may give rise to a different stress pattern in the region of the flaw which could alter the prestress benefit.

DISCUSSION AND SUMMARY OF OBSERVATIONS

From the foregoing presentation, the authors conclude that the initial proof test, if performed above NDT, is desirable in view of the beneficial stress redistribution that may result in the highly strained locations. Such a test is also necessary to demonstrate vessel integrity before it is placed in service. Further periodic proof tests above NDT are considered useful in disclosing critical flaws, even though brittle fracture is not likely to occur here. This conclusion is predicated on the uncertainty of nondestructive methods disclosing all flaws above the threshold of detection limits.

The majority of this section has been devoted to examining the benefits of warm prestressing, and of operation and proof testing below NDT, where catastrophic failure is possible. The authors have implied that enough uncertainty still exists to indicate that warm prestressing cannot be relied upon to guarantee safe operation below NDT. Four major factors lead to this conclusion.

1. Nuclear pressure vessel materials are strain rate sensitive, so that shock loading or a rapid burst of crack growth from a locally embrittled region could trigger failure. In this sense no static proof test guards against dynamic fracture initiation, and a dynamic proof test is required. However, such a test appears unrealistic from the practical viewpoint.

2. The possibility of aging as a means of crack tip embrittlement has been indicated for A302-B. This embrittlement could allow a rapid increase in crack growth and trigger a dynamic fracture initiation. Such failure possibilities appear likely only below NDT. Above this temperature the inherent dynamic toughness of the material is sufficient to absorb small local instabilities.

3. One cannot be certain that a vessel will be fracture safe below NDT even when operated less than the warm preload pressure. The test results on aging embrittlement reinforce this caution. Other work indicates that improper knowledge of imposed strain levels around flaws during the prestress may result in strains large enough to cause localized cracking and subsequent failure at loads below the preload level. The stress distribution in operation also may not be similar to that evidenced during the proof test. Such a possibility has never been explored. Consequently, other means of analysis are required to assure safe operation below NDT.

4. Considerable work remains to be done on flaw growth rates coupled with a nuclear environment. Other effects tending to increase flaw growth rates, such as corrosion and aging embrittlement, must be further examined before the magnitude of crack growth between proof tests can be safely predicted. The warm prestressing effect was shown to be of localized nature. It cannot be said with assurance how far a fatigue crack must extend during service to grow through this area of benefit and result in the brittle behavior that would be expected without the warm prestress.

Proof testing below NDT is not recommended, for most of the reasons summarized previously. The arguments for periodic testing and/or operation below NDT are not so strong that the procedure can be indiscriminately applied without carefully weighing the consequences. For the strain-rate-sensitive mild steels, the most positive fracture-safe course is to use these materials in the region considerably above the NDT, where high fracture toughness has been demonstrated.

SUMMARY

This report has discussed some of the environmental and operational factors that must be considered when evaluating pressure-vessel performance. Separate phenomena such as corrosion, fatigue, and irradiation embrittlement all have received considerable investigation. On the other hand, the possible synergistic effects of radiation in combination with other factors has received relatively little emphasis. The authors have considered two such effects, namely, the effect of applied stress coupled with irradiation and the effect of irradiation on low-cycle fatigue response.

The existing data relating to a coupling of stress and irradiation have been summarized and new data generated at NRL have been presented. None of the reported studies have revealed a significant effect of stress on the irradiation responses of carbon and low alloy steels or a synergistic effect of stress and radiation on mechanical properties performance. Stress during irradiation likewise does not appear to have a general effect on postirradiation annealing response.

The existing data relating to the effect of irradiation on fatigue life of reactor pressure vessel steels have been summarized. Tests of specimens fatigued after irradiation show a slight lowering of life in the low-cycle region and a slight raising of the high-cycle life. This result is commensurate with expected behavior of a higher strength steel. In-reactor fatigue investigations tentatively conclude that dynamic fatigue cycling during irradiation does not produce results grossly different from those obtained with irradiation followed by postexposure cycling. However, experimental tests to date have not examined flaw growth rates. This is an area to which future radiation-effects research must be directed.

The merits of proof testing as a means of ensuring safe vessel operation have been discussed. It is concluded that the initial proof test is of value, as are periodic proof tests when conducted above NDT. This statement is predicated on the uncertainty of nondestructive testing techniques in disclosing all potentially critical flaws. Certain problem areas present themselves when the proof test is used to assure operation in the "brittle region" below NDT. The correspondence of critical defect size with the lower limit of nondestructive tests appears to preclude proof testing below NDT without a real risk of failure. The prior use of warm prestressing in the ductile region above NDT may permit a safe proof test below NDT. However, new data presented by the authors indicate that the risk of aging embrittlement is a factor which may detract from the benefits of such an operation. The proof test is not a guarantee of safety from dynamic fracture initiation below NDT.

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13. ABSTRACT

The nuclear service performance of structural steels as influenced by static and cyclic stress applications during radiation exposure has been examined and documented with experimental results. The significance and merits of initial and subsequent proof tests of large structural components such as the hydro-testing of nuclear reactor pressure vessels have also been reviewed and evaluated. Performance following preload in the form of warm prestressing as well as aging embrittlement were among those factors considered.

A summary review of available data representing a coupling of static stress and radiation has suggested neither a significant effect of stress on irradiation responses, nor a synergistic effect of stress and radiation on mechanical properties performance. New data also indicate that stress during irradiation may not have a general influence on postirradiation annealing response. A parallel review of the cyclic stress behavior of pressure vessel steels did reveal a slight alteration of basic (preirradiation) low-cycle fatigue properties by neutron exposure. The general trend depicted was one of a radiation reduction of fatigue life at high strain amplitudes (low-cycle failure region) and of improvement at low strain amplitudes (high-cycle region). The trend was found consistent for both pre-post irradiation test methods and dynamic in-reactor testing.

Evaluations of the merits and objections for proof testing allowed a conclusion that both initial and periodic proof tests conducted above the nil-ductility transition temperature (NDT) are of definite value; however, operations below NDT either as a proof test or subsequent to warm prestressing above NDT were found objectionable.

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Notch ductility						
Radiation embrittlement						
Low cycle fatigue						
Postirradiation fatigue testing						

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