

The Effects of Warm Overstressing on Pressure Vessel Steel Properties

PVRC study shows that overstressing of flaws raises the fracture resistance of A516 Grade 70 and A533 Grade B but that subsequent strain aging causes some toughness reduction

BY L. N. SUCCOP, A. W. PENSE AND R. D. STOUT

ABSTRACT. The beneficial effects of warm prestressing have been demonstrated in a number of laboratory tests and, in principle, should have considerable value in enhancing the fracture capability of pressure vessels. However, there is some indication that the beneficial effects may be lost under actual vessel operational conditions such as with prolonged thermal exposure. During the warm prestressing (overstressing) process, plastic strains occur at regions in the vessel where high stress concentrations exist. When the material is subsequently exposed to elevated operating temperature, these regions become embrittled by the process of strain aging. This embrittlement may be enough to trigger a catastrophic failure in the pressure vessel. Hence, an investigation was undertaken to study the effects of strain aging embrittlement, arising from overstressing techniques, on two pressure vessel steels: A533B and A516 grade 70.

Charpy impact tests and tension tests, in conjunction with fracture toughness tests, were performed to determine the relative degree of material damage caused by strain aging. Different levels of strain and different cycles of straining and aging were studied. Different aging temperatures were also studied, and a stress relief treatment was investigated.

The results of the Charpy impact and tension test data indicated that strain aging embrittlement occurs in A516 grade 70 steel and A533B steel but is more severe for the A516 grade 70 than for the A533B. For both steels, repeated straining and aging treatments were noted to be more severe than single cycle strain age treatments. Different aging temperatures proved that 400° F had lesser effect on the A516 grade 70 steel than did 500 and 650° F. For the A533B steel aging at 400° F produced negligible

response, while 500 and 650° F aging produced more embrittlement. The stress relief treatment for both steels improved the Charpy impact properties.

The fracture toughness data for uniformly prestrained steels showed a moderate decreasing trend in K_{IC} for specimens that had been strained and aged. Specimens strained at a fatigue cracked notch were increased in toughness by the strain, but the benefits of this prestressing were degraded by subsequent aging treatments. High levels of notch strain produced K_{IC} values that were significantly higher than the base metal K_{IC} values. The K_{IC} values resulting from the high level notch strain followed by an aging treatment were decreased by the aging process. The K_{IC} values resulting from lower notch strain followed by an aging treatment, were not above the level of the base metal and were, in some cases, below it. Finally, both uniform and notch prestrain studies indicated that stress relief treatments, following straining and/or straining and aging treatments, tended to decrease the K_{IC} for both steels rather than improve it.

Introduction

This investigation concerns the strain aging embrittlement of pressure vessel steels as a result of overstressing techniques followed by prolonged thermal exposure. Since one of the primary requirements of pressure vessel operation is the prevention of unexpected catastrophic failure, vessels are carefully designed, fabricated, and nondestructively tested. Also, after the vessel has been fabricated it is required to undergo a proof test. This test usually consists of hydrostatic pressurization to 1.5 times the design pressure, as specified by Section VIII of the American Society of Mechanical Engineering (ASME) Boiler and Pressure Vessel Code. The hydrostatic test is necessary to provide evidence that the vessel will withstand operating pressure.

Such proof testing, however, is at stress levels high enough to produce plastic strain at regions in the vessel where mechanical stress concentrations exist, either because they are introduced by normal fabrication procedure or because vessel design incorporated such stress concentrations. In some cases this procedure, designed specifically to produce such strains, has been called prestressing or overstressing. In order to ensure against brittle fracture during the overstressing cycle, this is often done at temperatures above ambient and thus is called "warm prestressing."

Although warm prestressing and proof testing are similar insofar as executional features are concerned, the basic effect utilized and the end result desired are quite different.

In Yukawa's recent paper,¹ he defines warm prestressing as, "a procedure of subjecting the vessel to a pre-load or prestress under conditions where its fracture resistance is inherently high for the purpose of increasing the fracture resistance under subsequent conditions where the fracture resistance would otherwise be low. More specifically the procedure usually involves prestressing at some higher temperature to improve the fracture resistance at a lower temperature."

He defines proof testing as, "a procedure of subjecting the vessel to a stress level higher than the design or operating conditions for the purpose of ascertaining the lower bound of the ultimate fracture capacity. Alternatively, the purpose may be stated as ascertaining the absence of a flaw larger than a certain calculated size."

In both procedures, pressurization is the most feasible method of stressing the vessel. Using Fig. 1, Yukawa explains the two procedures in greater detail. The three fracture strength curves show schematically the influ-

L. N. SUCCOP is a former research assistant, A. W. PENSE is Associate Professor and R. D. STOUT is Dean of the Graduate School, Lehigh University, Bethlehem, Pa.

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ence of the two important factors of flaw size and temperature. For the kinds of steels used in pressure vessels, the general trends shown in Fig. 1 have been established by a large number of experimental studies.

The theory of warm prestressing assumes the existence of a "large" flaw. As shown in Fig. 1, the fracture strength associated with this large flaw is relatively low at some specific low temperature, T_1 . The objective of warm prestressing is to produce an increase in this low fracture strength by subjecting the structure to stress σ_2 at temperature T_2 or higher. At the same time care must be taken to perform the warm prestressing at conditions that ensure no possibility of fracture during the prestressing. The basic premise of this procedure is that after an excursion to stress σ_2 at T_2 or higher, the structure containing the large size flaw can ideally withstand any applied stress up to σ_2 at temperature T_1 . Even if this ideal amount of enhancement is not obtained, the premise is that some substantial increase in ultimate fracture capability is accomplished.

Nichols² points out that these benefits of warm prestressing arise in at least three ways, all of which are a consequence of local plastic deformation:

1. Work hardening the material at the tip of the crack, which leads to an increase in yield strength.
2. Producing local yield at the crack tip which introduces favorable

residual compressive stresses there when the structure is unloaded.

3. "Blunting" the crack tip which reduces its sharpness or severity.

As far as proof testing is concerned Yukawa explains that this procedure involves subjecting the vessel to stress level, σ_2 (Fig. 1) at some designated temperature such as T_1 , T_2 or T_3 . If the test is carried out at temperature T_1 and the vessel successfully withstands a stress level σ_2 , the absence of a "large" size flaw can be inferred. It does, however, leave open the possible existence of "medium" and "small" size flaws. If the proof testing to stress σ_2 is done at temperatures T_2 or T_3 , the absence of even a large flaw cannot be inferred. Also with this latter situation no information could be obtained about the ability to withstand stress σ_2 at temperature T_1 . Another consideration of proof testing is that the proof testing conditions, similar to the warm prestressing conditions, necessarily involve a compromise between being sufficiently discriminating for flaws and avoiding actual vessel failure.

In evaluating warm prestressing and proof testing procedures, Yukawa concludes that beneficial effects for warm prestressing have been demonstrated in a number of laboratory tests and, in principle, should have considerable value in enhancing the fracture capability of pressure vessels. On the other hand, proof testing is basically an inspection technique that provides only marginally small increments of

information. It should also be noted that proof testing, if done at higher temperatures, does at the same time constitute a warm prestress. Yukawa points out that the effectiveness of both procedures applies only to the extent that the sense and pattern of stressing reproduces those occurring in operational service, i.e., it is very difficult by pressure alone to attain stresses that match in magnitude and sense all of the operational stresses.

Both Yukawa and Nichols have considered some of the limitations of the procedures arising from unknown factors. These factors are:

1. The strain aging sensitivity of current pressure vessel steels.
2. The effect of prolonged thermal and radiation exposure on the retained effectiveness of warm prestressing.
3. The response to prestressing of radiation embrittled steels. The latter two factors obviously apply to the use of vessels for nuclear reactor service, but the possibility of strain aging will apply to all vessels that are strained and subsequently placed in elevated temperature service.

As indicated above, the plastic strain produced during the overstressing combined with subsequent elevated temperature exposure can lead to strain aging embrittlement, particularly at the stress concentration locations in a vessel. This embrittlement is potentially dangerous since such locations will always be at higher than working stress and could conceivably trigger a failure at pressures below the

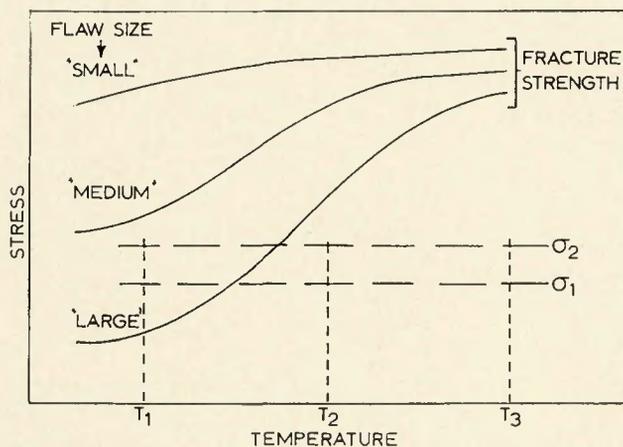


Fig. 1—Schematic representation of proof testing and warm prestressing

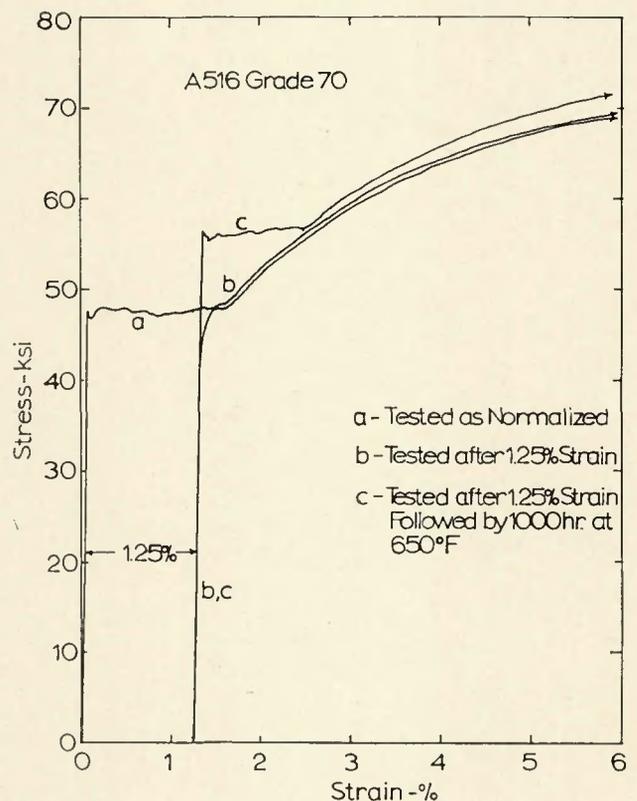


Fig. 2 (right)—The response of A516 Grade 70 to strain aging

Table 1—Chemical Analysis of the Steels, %

| Steel | Grade | C | Mn | P | S | Si | Ni | Cr | Mo | Cu | Al |
|-------|-------|-----|------|------|------|-----|-----|-----|-----|-----|------|
| A533 | B | .18 | 1.25 | .024 | .025 | .24 | .52 | .13 | .51 | .29 | .024 |
| A516 | 70 | .23 | .97 | .010 | .020 | .25 | .09 | .05 | .02 | .25 | .025 |

proof stress level, thus eliminating the beneficial effects of prestressing.

Basically, strain aging is an embrittling phenomenon that takes place after or during plastic strain. Baird³ reports that the strain aging process consists of two parts:

1. The formation of solute "atmospheres" around dislocations produced by straining.

2. The formation of rod-like precipitate particles centered on the dislocation core.

At the present time, strain aging is considered to be due primarily to the migration of carbon and nitrogen atoms to dislocations and locking them. Other elements are also important and Baird³ summarizes all elements affecting strain aging into two groups:

1. The solutes which can lock dislocations and which can diffuse sufficiently quickly to produce strain aging (C and N).

2. The elements which themselves do not produce strain aging but affect the process by altering the concentration or mobility of the solute atoms that produce strain aging. These elements can themselves be divided into

four categories:

(a) Elements which interact weakly or not at all with nitrogen and carbon—Cu, Ni, Mn, P.

(b) Nitride formers—Al, Si, B.

(c) Carbide formers—Mo.

(d) Nitride and carbide formers—Cr, V, Nb, Ti.

With respect to the alloying elements listed above, it becomes apparent that killed steels containing aluminum and/or those containing strong carbide (and nitride) formers such as Ti, V, Nb and to a lesser extent Cr, will be less susceptible to strain aging phenomena, while semikilled steels may be more so. Previous PVRC investigations⁴ of strain aging have shown this to be generally true. Experimentally, the general effect of strain aging on the tensile properties of A516 grade 70 is illustrated in Fig. 2. If the specimen is strained to a point well into or beyond the lower yield extension, unloaded and then immediately retested, the stress-strain curve follows the same curve. However, if the specimen is unloaded after plastic strain and then allowed to age at room temperature or above, the

discontinuous yielding behavior returns and the stress strain curve follows a curve such as (c) in Fig. 2. The yield point is now higher than the flow stress at the end of prestressing. This increase in yield or flow stress on unloading and aging is the most universal indication of strain aging. There may also be an increase in ultimate tensile strength and a decrease in elongation and reduction of area, but these do not always take place.

The effect of strain aging on the tensile and impact properties of steels has been extensively studied by PVRC⁴ and others.⁵ The PVRC investigations have shown that both carbon and alloy pressure vessel steels have a tendency for strain aging in the 500–700° F range if given a tensile prestrain of 5%. For the carbon steels, the maximum aging response occurred at 500° F while for alloy steels it was closer to 700° F. The aging was characterized by an increase in yield and, to a lesser extent, tensile strength of the steel and a decrease in tensile ductility. The Charpy impact transition temperature increased substantially as a result of aging. The effect of strain aging on the fracture toughness of carbon and low alloy steels that were warm prestressed, then aged, has been studied by Harrison and Fearnough⁵ and by Hawthorne and Loss.⁶ These results show that although high levels of tensile prestrain raised the toughness of the specimens, a 25% loss of the

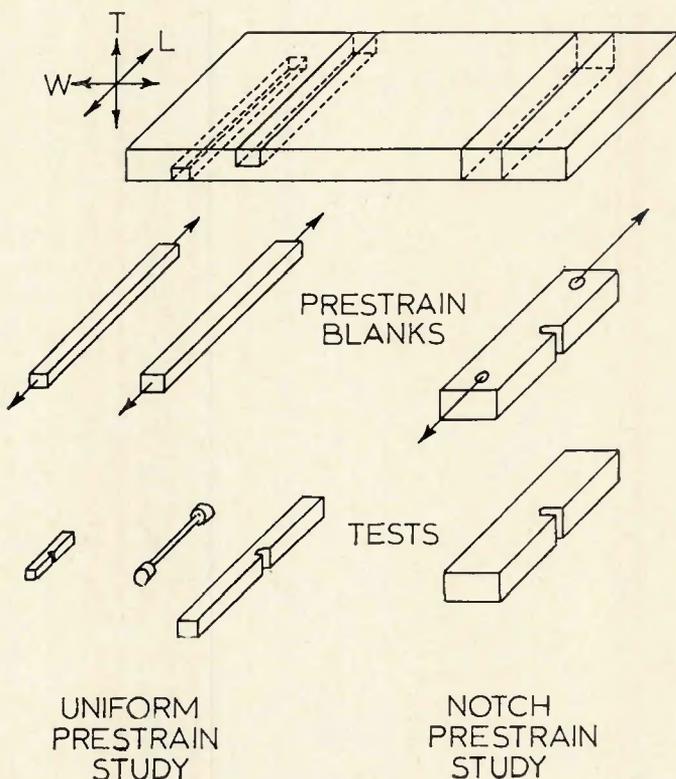


Fig. 3—Specimen orientations in A516 Grade 70 plate

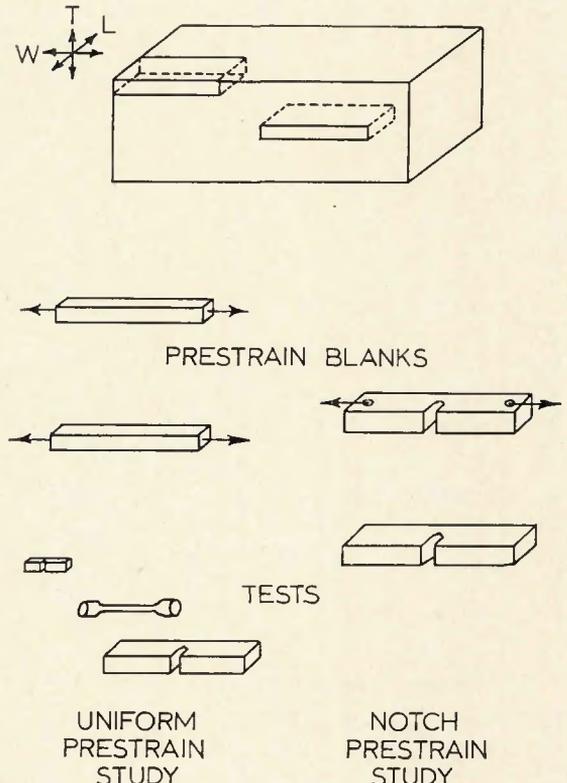


Fig. 4—Specimen orientations in A533 Grade B plate

Table 2—Tensile Test Data

| A516 grade 70 steel | | | | Base properties | A533B steel (heavy section) | | | |
|------------------------|--------------------------------|---------------|----------------------|--|-----------------------------|--------------------------------|---------------|----------------------|
| 0.2% offset yield, ksi | Ultimate tensile strength, ksi | Elongation, % | Reduction in area, % | | 0.2% offset yield, ksi | Ultimate tensile strength, ksi | Elongation, % | Reduction in Area, % |
| 48.4 | 76.6 | 34.0 | 66.6 | | 72.5 | 90.7 | 24.5 | 57.1 |
| 51.5 | 78.1 | 31.0 | 64.5 | Strained 1.25% at 200° F | 75.1 | 90.6 | 24.9 | 56.4 |
| 56.9 | 80.7 | 33.5 | 64.8 | Strained 1.25% at 200° F-aged at 650° F for 1000 hrs 1 cycle | 79.2 | 89.9 | 19.4 | 52.4 |
| 67.5 | 84.3 | 27.0 | 61.0 | Strained 1.25% at 200° F-aged at 650° F for 1000 hr 3 cycle | 90.5 | 96.9 | 17.0 | 53.7 |
| 55.6 | 79.9 | 31.7 | 67.5 | Strained 1.25% at 200° F then aged at 500° F for 1000 hr 1 cycle | 79.8 | 89.9 | 19.5 | 55.2 |
| 85.6 | 91.2 | 24.2 | 61.3 | Strained 1.25% at 200° F then aged at 500° F for 1000 hr 3 cycle | 92.4 | 97.2 | 18.5 | 59.3 |
| 58.4 | 80.0 | 31.3 | 65.6 | Strained 1.25% at 200° F then aged at 400° F for 1000 hr 1 cycle | 77.9 | 86.2 | 24.2 | 58.3 |

Table 3—Charpy V-Notch Impact Test Data

| A516 grade 70 steel | | | | | Condition | A533B steel (heavy section) | | | | |
|--|----------|---------|-----------|---------------------|---|--|----------|---------|-----------|---------------------|
| Transition temperature (°F) at 50% shear | | | | Shelf energy, ft-lb | | Transition temperature (°F) at 50% shear | | | | Shelf energy, ft-lb |
| 15 ft-lb | 20 ft-lb | 15 mils | 50% shear | | | 15 ft-lb | 30 ft-lb | 15 mils | 50% shear | |
| #1 -83 | -68 | -85 | +30 | 83 | Base properties | | | | | |
| #2 -74 | -57 | -89 | +33 | 81.5 | | | | | | |
| #3 -84 | -63 | -90 | +35 | 77 | | | | | | |
| -60 | -41 | -68 | +46 | 82.5 | Strained 1.25% at 200° F | -14 | +12 | -20 | +60 | 46 |
| -45 | -33 | -46 | +56 | 65 | Strained 1.25% at 200° F then aged at 650° F for 1000 hr 1 cycle | -11 | +64 | 0 | +64 | 43 |
| +2 | +20 | -6 | +56 | 72 | Strained 1.25% at 200° F then aged at 650° F for 1000 hr 3 cycles | +20 | +87 | +6 | +68 | 40 |
| -47 | -32 | -45 | +43 | 76 | Strained 1.25% at 200° F then aged at 500° F for 1000 hr 1 cycle | +40 | +116 | +30 | +77 | 36 |
| +16 | +37 | +20 | +69 | 62 | Strained 1.25% at 200° F then aged at 500° F for 1000 hr 3 cycles | +28 | +75 | -4 | +58 | 40 |
| -50 | -31 | -46 | +48 | 72 | Strained 1.25% at 200° F then aged at 400° F for 1000 hr 1 cycle | +20 | +100 | +22 | +76 | 48 |
| -43 | -28 | -51 | +56 | 85 | Strained 1.25% at 200° F then aged at 650° F for 1000 hr 1 cycle stress relief at 1100° F for 8 hr furnace cool 50° F/hr | -14 | +52 | -40 | +56 | 52 |
| -33 | -18 | -36 | +60 | 85 | Strained 1.25% at 200° F then aged at 650° F for 1000 hr 3 cycles stress relief at 1100° F for 8 hr furnace cool 50° F/hr | -38 | +25 | -51 | +48 | 63 |
| | | | | | | -36 | +33 | -36 | +62 | 61 |

prestressing effect resulted from a 500 hr age at 550° F. When tensile and then compressive prestrain is applied the toughness is not improved, and subsequent aging at 482° F for 1/2 hr reduces the toughness below the original plate level.

As may be seen from the above discussion, the use of warm prestressing has many advantages. It will accomplish not only a proof test but will also serve as a means of brittle fracture control. Indeed, in cases where thermal stress relieving of a completed vessel is not economical or practical, it has been suggested² that mechanical overstressing should be used in place of thermal stress relief. Even in cases where overstressing does not replace thermal-stress relief, traditional proof testing at levels high enough to produce plastic yielding will result in the same effect in terms of subsequent aging behavior. When a pressure vessel is placed in service where

reliability is particularly critical, and flaw growth is known to be likely during operation, periodic proof testing of the vessel has sometimes been part of the operating plan. Thus repeated cycles of plastic straining and service are anticipated for such vessels.

The factors that counterbalance the advantages are metallurgical ones—the response of the material to radiation damage and to aging phenomena induced by the prestraining. If either are severe, the potential advantages of prestraining will be either negated or perhaps overshadowed by the potential danger of reduced fracture resistance.

Since the metallurgical response of pressure vessel steels to warm prestraining in the presence of flaws has not been given sufficient attention to clarify the extent to which material degradation occurs under conditions of prestraining and aging, the Pressure

Vessel Steels Subcommittee, Materials Division, PVRC, undertook an experimental program at Lehigh University to examine this response in several pressure vessel steels. The objectives of the program were to determine what tensile prestrain followed by representative elevated temperature (aging) cycles would do to the material itself, and also what influence prestrain and aging would have on the fracture behavior of the same material with a controlled flaw present during the prestrain. As a corollary study, the influence of thermal stress relief on strain aged material, with and without flaws, was also examined.

Method and Procedure

The two steels investigated in this project were A533B and A516 grade 70. Table 1 shows the chemical compositions of these materials.

The A533B steel was sectioned into

Table 4—Summary of Experimental Program

| | | | | |
|----------------|----------------------|---|---|---|
| Base condition | Uniform strain tests | Strained 1.25% at 200° F | Aged 1000 hr at 400° F | |
| | | | Aged 1000 hr at 500° F | Strained 1.25% at 200° F, aged 1000 hr at 500° F plus strained 1.25% at 200° F, aged 1000 hr at 500° F |
| | | | Aged 1000 hr at 650° F ^a | Strained 1.25% at 200° F, aged 1000 hr at 650° F plus strained 1.25% at 200° F, aged 1000 hr at 650° F ^a |
| | Notch strain tests | Low notch strain at 200° F ^{a, b} | Aged 400 hr at 650° F ^a | |
| | | | Aged 1000 hr at 650° F ^a | Low notch strain at 200° F, aged 1000 hr at 650° F plus low notch strain at 200° F, aged 1000 hr at 650° F ^a |
| | | Medium notch strain at 200° F ^{a, c} | Aged 1000 hr at 650° F | |
| | | | High notch strain at 200° F ^{a, d} | Aged 1000 hr at 650° F |

^a Data obtained both before and after thermal stress relief, 8–12 hr at 1100° F.
^b Prestrain K of 38 ksi $\sqrt{\text{in.}}$ for A516 grade 70, 40 ksi in. for A533B—5% plastic zone size.
^c Prestrain K of 63 ksi $\sqrt{\text{in.}}$ for A533B—15% plastic zone size.
^d Prestrain K of 88 ksi $\sqrt{\text{in.}}$ for A516 grade 70, 77 ksi $\sqrt{\text{in.}}$ for A533B—25% plastic zone size

$3/4$ – $1\frac{1}{2}$ in. thick slices from the $7\frac{1}{2}$ in. thick plate as shown in Fig. 3. The slices were then given a heat treatment to match the thermal history of the center of the quenched, tempered and stress relieved plate 6 in. thick.⁷

The A516 grade 70 steel (2 in thick) needed no additional heat treatment and was tested in the normalized condition.

Standard Charpy impact tests and tensile tests were then performed on both steels to determine the base mechanical properties. These results are shown in Tables 2 and 3. For the first material, A533B, the $1\text{--}1\frac{1}{4}$ in. slices were of such a shape that the specimens had to be cut transverse to the rolling direction (as shown in Fig.

3) in order to save material. For the A516 grade 70 steel, the specimens were cut parallel to the rolling direction (shown in Fig. 4) from center, surface and quarter thickness positions of the 2 in. thick plate. Mechanical tests, as shown in Table 3 indicated that the A516 grade 70 plate was relatively mechanically uniform through its thickness. Hence, it was decided that specimens cut from any of the three positions would give satisfactory evaluation of the mechanical properties of a 2 in. thick plate.

After the base mechanical properties of the steels were determined, uniform and notch prestrain specimen blanks were machined from the plates as shown in Figs. 3 and 4. The straining and testing schedule for the speci-

mens is listed in Table 4.

Uniform Prestrain Study

For the uniform strain part of the study the blanks, ranging in size from $5/8 \times 1\frac{1}{2} \times 18$ in. to $1\frac{1}{4} \times 1\frac{3}{4} \times 27$ in., were uniformly strained 1.2–1.5% at 200° F in a standard universal testing machine. An environmental testing chamber, using forced air, was used to heat the specimens. The deformation during the straining was monitored by a dial gage extensometer attached to the outside of the chamber.

Following the straining, some of the blanks were immediately machined into Charpy and tensile specimens, while others were aged for 1000 hr at 650, 500, and 400° F. After the single strain-age cycle, some of the bars aged at 650° F were machined into standard Charpy, tensile, and slow bend fracture toughness specimens. The others were then given two more strain-age cycles before they were made into the same three types of specimens. For materials aged at 500 and 400° F, only Charpy and tensile specimens were investigated. The 500 and 650° F aging temperatures alone were investigated for the three-cycle-strain-age condition.

The Charpy and tension specimens were tested on a standard impact testing machine and a standard 10,000 lb capacity universal testing machine, respectively. Tension data for A516 grade 70 steel over a range of temperatures from –250 to +200° F were also obtained.

The slow bend fracture toughness specimens for the uniform strain part of the study, seen in Fig. 5, ranged in size from $0.60 \times 1.20 \times 10$ in. to $0.8 \times 1.50 \times 10$ in. After being cut from the blanks and machined they were then precracked at 286–300 cycles/sec.

A three point bend loading arrangement was used to grow fatigue cracks that varied from 0.45 to .055 times the depth of the specimen, W . The chevron notch configuration, shown in Fig. 5 was used in order to form a fatigue crack which did not deviate from the notch plane and yet extended substantially beyond the notch root throughout most of the specimen thickness. This substantial extension was necessary to avoid undue influence of the notch on the crack stress field. The chevron notch also expedited nucleation of the fatigue crack.⁸ In all the specimens, the cracks were grown in two steps. The first step grew the crack beyond the notch root at a fast growth rate. The second step grew the crack at a slow rate so that the final .050 in. was grown in 50,000 cycles or more.

Following precracking, all the slow bend specimens were loaded in a three

$$K_Q = \frac{P_Q L}{B W^{3/2}} f\left[\frac{a}{W}\right]$$

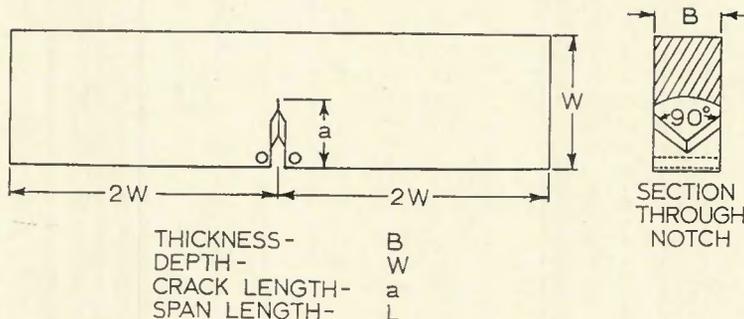


Fig. 5—Slow bend specimen

point bend fixture, at approximately a 60–120 ksi $\sqrt{\text{in.}}$ per min loading rate — Fig. 6. A double cantilever clip-in displacement type gage was inserted into attachable knife edges already on the specimen. The displacement output from the clip gage, along with the load output from the load cell, was fed into an x-y recorder. Different test temperatures from -250 to -150° F were obtained with liquid nitrogen applied through perforated copper coils surrounding the specimen in a cooling box insulated with 2 in. thick styrofoam. The temperature was monitored by means of a thermocouple attached to the center thickness of the specimen. The specimen was held at temperature for 15 min before testing, and a minimum of 2 specimens were tested for each condition.

Notch Prestrain Study

In developing a test program to study warm prestressing, it soon became apparent that uniform straining of a small flaw free specimen could not duplicate the mechanical effects that exist in the region at the tip of a sharp crack in a thick walled pressure vessel. The stress state conditions represented by this situation may, indeed, never be duplicated satisfactorily in any laboratory test. However, a simulation of this condition could be attempted by studying the behavior of a specimen which contained a sharp natural flaw. The influence of prestrain and aging on this specimen, in conjunction with more traditional tests, could then be used to clarify the relative role of those effects, such as response to aging and work hardening, which are primarily metallurgical in nature and those, such as crack blunting and residual stress state, which are primarily mechanical in origin.

The test specimens used to duplicate such a condition were the slow

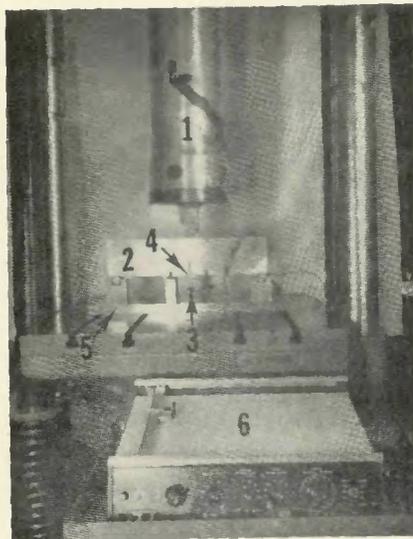


Fig. 6—Slow bend test apparatus: 1—load cell; 2—specimen; 3—clip gage; 4—knife edges; 5—bend fixture; 6—X-Y recorder

bend and the compact tension fracture toughness specimens. The basic procedure was to machine and fatigue crack these specimens as in the uniform strain study, and then to treat them as a section from the membrane of a vessel given a prestressing treatment. This was accomplished by giving a tensile prestrain to the specimen and then applying the aging and/or stress relief treatments which were to be studied. The tensile prestrain created a plastic zone at the tip of the fatigue crack which simulated, at least to some degree, that associated with a flaw in a vessel.

Three levels of tensile prestrain were employed in the investigation. The first, representing a condition of a small flaw in the wall of a thick vessel, was applied at a prestrain only great enough to produce a relatively low stress intensity at the fatigue crack tip. A plastic zone of small size, about

5% of the remaining uncracked section of the specimen, resulted. The second level of tensile prestrain was higher, designed to simulate the behavior of a larger flaw in a vessel. This produced a plastic zone size of about 15% and was used for A533 grade B only. The third level, representing a high stress intensity such as might occur at the tip of a large deep flaw, produced a plastic zone size of about 25% of the remaining section.

For the notch prestrain part of the study, the slow bend specimens ranged in size from 1.0 x 2.0 x 18.0 in. for A533B to 2.0 x 4.0 x 18.0 in. for A516 Grade 70. After being cut from the plates and machined as shown in Figs. 4 and 5, they were precracked, as described before, on a fatigue testing machine. All the specimens, in the configuration shown in Fig. 8 were heated to 200° F, slowly loaded in tension to a stress level based on the specimen size, and held for five minutes. As described above, prestress levels of 38 and 88 ksi $\sqrt{\text{in.}}$ were studied for A516 grade 70, while 40, 63 and 77 ksi $\sqrt{\text{in.}}$ were used for A533 B. The prestrain K level was calculated on the basis of the preload and the crack length measured on the surface of the specimen after fatigue cracking. The formula used was that recommended for single-edged cracked plates in tension in ASTM STP 410:⁹

$$K(\text{applied}) = Y \cdot \frac{Pa^{1/2}}{BW} \quad (1)$$

where K = applied stress intensity factor ksi $\sqrt{\text{in.}}$; P = preload (lbs.); a = crack length after precracking (in.); B = thickness (in.); W = width (in.); Y = calibration factor for specimen geometry.

The corresponding plastic zone sizes (r_y) were calculated from:

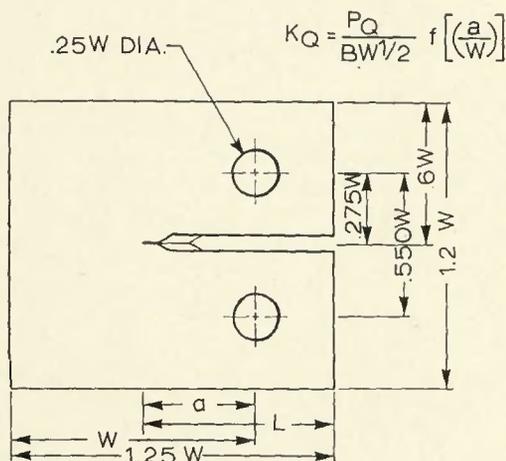


Fig. 7—Compact tension specimen

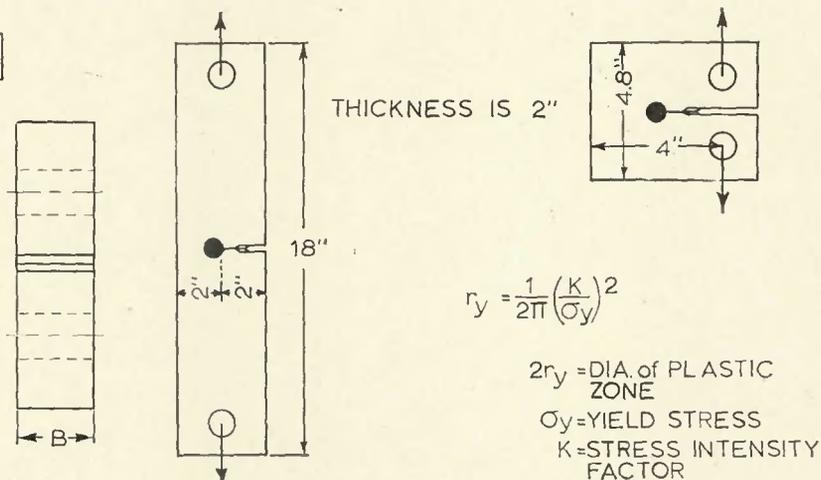


Fig. 8—Prestraining technique for the specimens

$$r_y = \frac{1}{2\pi} \left(\frac{K}{\sigma_y} \right)^2$$

$2r_y$ = DIA. of PLASTIC ZONE

σ_y = YIELD STRESS

K = STRESS INTENSITY FACTOR

Table 5—Low Temperature Tension Test Data (−250° F to +200° F)

| A516 grade 70 steel | | | | A533B steel ^a | | | | |
|------------------------|--------------------------------|---------------|----------------------|--------------------------|------------------------|--------------------------------|---------------|----------------------|
| 0.2% offset yield, ksi | Ultimate tensile strength, ksi | Elongation, % | Reduction in area, % | Temperature, °F | 0.2% offset yield, ksi | Ultimate tensile strength, ksi | Elongation, % | Reduction in area, % |
| 96.6 | 118.0 | 35.3 | 53.6 | −250° F | 114.1 | 124.5 | 21.0 | 56.7 |
| 68.0 | 98.7 | 39.0 | 55.7 | −150° F | 91.6 | 111.4 | 20.3 | 60.7 |
| 56.9 | 88.5 | 37.0 | 61.8 | −50° F | — | — | — | — |
| 54.0 | 87.1 | 36.2 | 61.2 | 0° F | — | — | — | — |
| 48.4 | 76.6 | 35.0 | 66.6 | +70° F | 71.4 | 91.4 | 18.0 | 65.5 |
| 47.5 | 75.0 | 36.2 | 68.7 | +200° F | 67.6 | 86.6 | 17.0 | 67.5 |

^a From Mager and Thomas.¹⁰

$$r_y = \frac{1}{2\pi} \left(\frac{K_I}{\sigma_y} \right)^2 \quad (2)$$

Following the prestress, some of the specimens were plane strain fracture toughness tested as described before and the others were aged for 400 and 1000 hr at 650° F. After the notch prestrain–1000 hr age cycle some of the specimens were tested while the remaining ones were given two more notch prestrain–age cycles of 1000 hr aging each before they were finally tested. The prestrain *K* level used during the second and third cycle was the same as applied previously. As before, a minimum of two specimens were tested for each condition.

After all the specimens had been machined and after a large number had been tested, it was decided that some additional conditions, particularly stress relief treatments, should be investigated. Since most of the test material available had now been used for the slow bend tests, in order to undertake these tests it was necessary to modify the specimen design in order to conserve material. Therefore, additional fracture toughness specimens were fabricated of the compact tension type, shown in Fig. 7, from

the broken ends of the slow bend fracture toughness specimens.

Only specimens tested in the base condition (without prestrain or aging) were used for specimen blanks and the portions near the fracture surfaces were not utilized. These specimens were 2.0 x 1.0 x 2.087 in. for A533B and 4.0 x 2.0 x 4.175 in. for A516 grade 70. The specimens were prestrained as shown in Fig. 7 and were tested in tension. Data on the condition of stress relief after prestrain and after aging was obtained by fabricating specimens, giving them the prior treatment to be studied, and then stress relieving them at 1100° F for 8 or 12 hr. The stress relief was followed by a furnace cool (approximately 50° F/hr). Both Charpy impact tests and fracture toughness tests were performed on these specimens. The A533B specimens were stress relieved 12 hr, the A516 grade 70, 8 hr.

Some special comments need to be made with regard to the validity of the fracture toughness test data. In one sense, almost all of the fracture toughness test data may be regarded as valid *K_{IC}* information. Only specimens which received very high prestrain values and thus had high values

of *K_{IC}* on testing were invalid by specimen size criteria. All load-displacement records were linear or nearly so and all specimens failed before a 5% secant modulus line was reached.⁸ Indeed, low testing temperatures were employed to ensure that this would be so.

In another sense, none of the notch prestrained specimens could be considered as “valid.” This is because, after careful precracking in fatigue, all of the specimens except the base condition ones were deliberately given treatments that were designed to simulate strained flaws in vessels. As a result, each was given a strain that would introduce a substantial but controlled plastic zone at the crack tip. Because of this large plastic zone, none of these tests could be considered as normal fracture toughness tests.

Results and Discussion

Charpy V-Notch Impact Test Data

The results of the Charpy V-notch impact tests are presented in Table 3. It should always be kept in mind when viewing these data that the A533B has been given a heavy section heat treatment simulating a 6 in. thick plate, while the A516 grade 70 steel is 2 in thick.

The base properties of the A533B steel indicate that this material has generally less impact toughness than the A516 grade 70 steel and a low upper shelf energy. However, this result is expected from the heavy section heat treatment given to the A533B plates and the fact that they represent transverse not longitudinal properties. The plate used in this program was not made to A533B specifications but was quenched and tempered nickel modified A302 grade B steel. It prob-

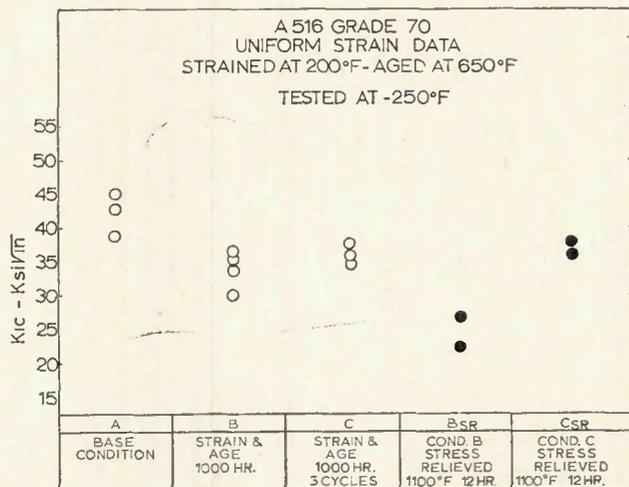


Fig. 9—Influence of uniform straining, aging and stress relief on the fracture toughness of A516 Grade 70 steel

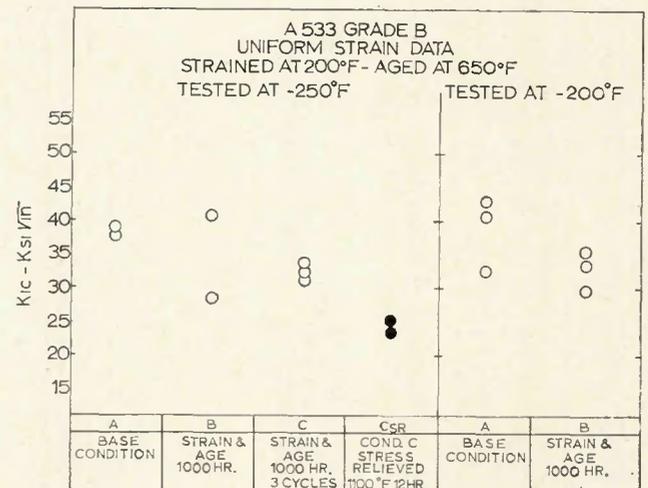


Fig. 10—Influence of uniform straining, aging and stress relief on the fracture toughness of A533 Grade B steel

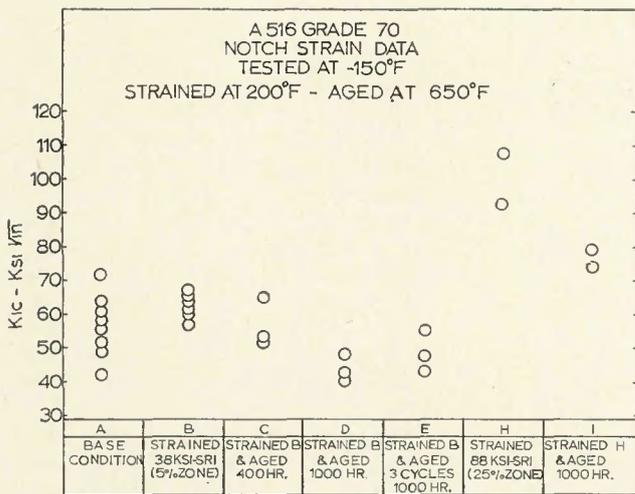


Fig. 11—Influence of notch straining and aging on the fracture toughness of A516 Grade 70 steel

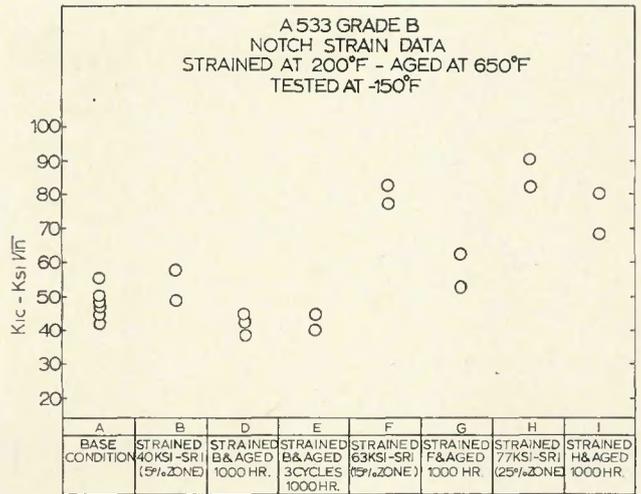


Fig. 12—Influence of notch straining and aging on the fracture toughness of A533 Grade B steel

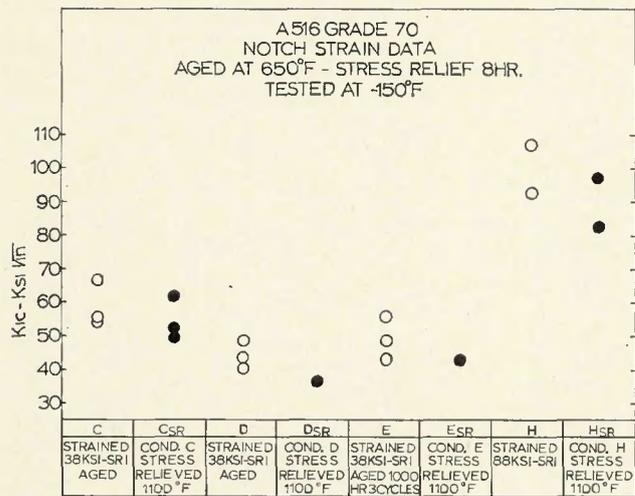


Fig. 13—The influence of notch strain, aging, and stress relief on the fracture toughness of A516 Grade 70

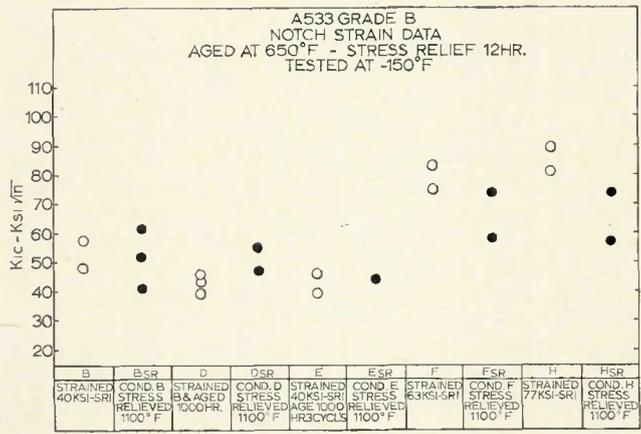


Fig. 14—The influence of notch strain, aging, and stress relief on the fracture toughness of A533 Grade B

ably does not wholly represent the melting or manufacturing practice for current A533 B heats.

The as-strained condition for the A516 grade 70 steel showed an increase of approximately 20° F for both the 15 ft-lb and 20 ft-lb transition temperatures as well as for the 15 mil (lateral expansion) transition temperature. The 50% fracture appearance transition temperature increased slightly and the shelf energy also increased slightly compared to the base properties. The A533B material showed a similar response to this condition, except to a lesser degree. It may be seen that the 30 ft-lb transition temperature increased 52° F. This response is regarded as large; however, it may also be noted that the shelf energy decreased slightly which shifts the upper end of the data curve to lower values and thus contributes to a larger change in the higher energy transition temperatures.

When both steels were strained and aged 1 cycle at 650° F, the transition

temperatures increased further and the shelf energies decreased. For the A516 grade 70 steel the 15 ft-lb, 20 ft-lb, and 15 mil transition temperatures increased approximately 35° F above the base property level. The 50% fracture appearance transition temperature increased 20° F above its base level and the shelf energy was noted to fall 10–15% below its original level. For the A533B steel the 15 ft-lb and 15 mil transition temperatures increased approximately 30° F above their base levels while the 30 ft-lb transition was observed to increase 75° F above its base level (recall that 52° F of this increase was noted in the as-strained condition). The 50% fracture appearance transition for this steel increased only 8° F above its original level while the shelf energy decreased about 12%.

Upon straining and aging both materials 3 cycles at 650° F, substantial increases in transition temperatures were noted for the A516 grade 70 steel while lesser increases were ob-

served for the A533B steel. For the A516 grade 70 steel, the 15 ft-lb, 20 ft-lb and 15 mil transitions increased approximately 50° F over the single cycle strain age level. The 50% fracture appearance transition remained the same and the shelf energy slightly increased to 72 ft-lb. For the A533B steel the 15 ft-lb, 30 ft-lb, and 15 mil transitions increased about 25° F over the single cycle strain age level. The 50% fracture appearance transition increased slightly to 77° F, while the shelf energy decreased to 36 ft-lb. The behavior exhibited by both steels indicates that repeated straining and aging is more severe than a single cycle treatment.

As far as aging temperature is concerned, 400° F produced lesser aging in the A516 grade 70 steel and none at all in the A533B. For the 500 and 650° F temperatures, little difference in respective embrittlement was noted.

The results of the stress relief treatment showed general improvement in

toughness over the aged condition. For the A516 grade 70 steel the single cycle strain age treatment followed by the stress relief treatment showed no improvement (no decrease) in the material transition temperatures, but the shelf energy increased back to its base level. However, for the repeated straining and aging condition followed by the stress relief treatment, the results indicate a substantial improvement in the transition temperatures. For the A533B steel similar behavior was noted, indeed, the stress relief treatment improved the properties to the degree that they were better than the base condition. For both steels these results indicate that stress relieving results in improved toughness over the aged condition. It should also be noted that after stress relieving treatments were given to the 1-cycle and 3-cycle strain aging conditions, the transition temperatures of both steels were approximately the same despite the original difference in base properties. The A516 grade 70 steel, however, still maintained a higher shelf energy.

Tensile Test Data

A summary of the tensile test data for the base conditions and the different straining and aging cycles is presented in Table 3. In addition, tensile data over a range of temperatures for both steels are presented in Table 5.

For both steels, the yield strength, tensile strength, elongation, and reduction of area data are typical of the changes associated with strain aging.³ As explained in the Introduction, strain aging increases the yield strength while the tensile strength, elongation, and reduction of area are less affected. For the A516 grade 70 steel the increase in yield strength from the base condition to the strain age 3-cycle condition was approximately 75%. For the A533B steel, the same conditions showed about a 28% increase. This result is in agreement with the Charpy impact test data and once again suggests that the A533B steel is less susceptible to strain aging. Based on yield point elevation, it appears that the repeated straining and aging condition produced the most embrittlement. This result is also in agreement with the impact data.

The effect of different aging temperatures was consistent with the impact data. The 500 and 650° F aging temperatures produced no significant variation in respective strain aging behavior. The 400° F aging temperature produced lesser aging in the A516 grade 70 steel than did the other temperatures, and hardly any

aging at all in the A533B material.

The tensile data shown in Table 5 are results that were obtained for use in validating fracture toughness tests. The A533B data were obtained from another investigation¹⁰ and are representative of this steel heat treated to the yield and tensile strength used in this investigation.

Fracture Toughness Test Data

The results of the fracture toughness tests are shown in Figs. 9–14. It should always be kept in mind when viewing these data that both the uniform and notch prestraining was performed at 200° F while the fracture testing was conducted at or below 150° F.

Uniform Prestrain and Aging Results

The uniform prestrain K_{IC} results for the A533B and A516 grade 70 steels are shown graphically in Figs. 9 and 10. The steels showed trends for a decrease in K_{IC} for increasing amounts of straining and aging. For the A533B the single cycle strained and aged condition reduced the average toughness approximately 10%. For the A516 grade 70 steel, the same treatment lowered the average K_{IC} value about 20%. For the triple cycle straining and aging condition, the A533B indicated a further drop to a value 16% below the original plate, while the A516 grade 70 steel showed a slight increase in toughness. This increase is small and the magnitude of changes in toughness between the 1-cycle and 3-cycle treatments may have been the result of test scatter. It should be noted for both steels that the repeated straining and aging treatment did not affect K_{IC} as significantly as it did Charpy impact and tensile results. The K_{IC} test results indicate that the A533B steel is not as susceptible to strain aging as the A516 grade 70 steel, which is consistent with the impact test data.

Notch Prestrain and Aging Results

The notch prestrain and aging results for both steels are found in Figs. 11–14. For both steels, the base condition tests show extensive scatter, and thus interpretation of the aging results is less certain, particularly at the lower prestrain levels. However, the influence of the lower level prestrain (38–40 ksi $\sqrt{\text{in.}}$) did appear to produce a slight improvement in fracture toughness in the specimens which was subsequently lost by aging. Since the plastic zone produced by the prestrain was small, the toughness improvement was limited. It should be remembered that, although the prestrain loads were less than the observed K_{IC} , the prestrain was applied

at 200° F. As a result, the plastic zone resulting from the prestrain is the same as or greater than that at failure.

Beyond 1000 hr of aging, repeated straining and aging did not produce a continued loss in toughness. Within the limitations imposed by test scatter it appears that the final condition of aging was to below the original toughness level of the plate. These results confirm the uniform strain test data, but also demonstrate that the loss in toughness in the notch prestrained condition is less than in the uniform tests.

The results of tests on material tested in the condition of higher levels of notch prestrain (63, 77 and 88 ksi $\sqrt{\text{in.}}$) followed by aging are also shown on Figs. 11 and 12. It is clear that the higher levels of warm pre-stressing did produce improvement in fracture toughness, but that this toughness improvement is to a large extent lost through subsequent aging. The aging loss left a residual improvement in toughness over the base condition of from 20 to 65%, as compared to almost 100% increase produced by the warm pre-stressing alone. These experimental data confirm the results of Hawthorne and Loss,⁶ and further suggest that strain aging can negate much of the benefit of even high levels of prestrain.

Higher levels of prestrain were not employed in this study and higher levels are probably not practical in actual practice. At the level 88 ksi $\sqrt{\text{in.}}$ slow crack growth occurred in A516 grade 70 during prestraining. Thus the A516 Grade 70 "H" specimens reported in Fig. 11 have already reached the limit on experimental prestrain values without risk of failure during prestraining.

Thermal Stress Relief Results

The influence of thermal stress relief on both the uniformly strained and notch prestrained material produced results that were both informative and puzzling. These results are seen on Figs. 9, 10, 13 and 14 where a variety of strained and aged conditions are plotted along with the equivalent material given a thermal stress relief of from 8 to 12 hr at 1100° F. It may be seen by such a comparison that, in almost every case, thermal stress relief was not helpful in improving toughness of either notch or uniformly prestrained steel and served to decrease toughness when applied directly to notched specimens after prestrain and prior to aging. The fracture surfaces of the highly prestrained material before and after thermal stress relief were suggestive of this toughness decrease as may be seen in

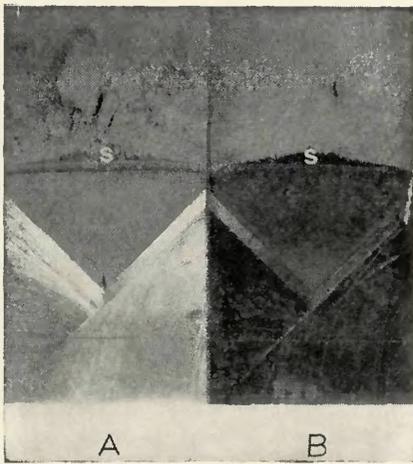


Fig. 15—A516 Grade 70 strained 88 ksi $\sqrt{\text{in.}}$ at 200° F and tested at -150° F. A—no stress relief B—stress relieved. S—denotes slow crack growth during straining

Figure 15.

Since the influence of prestrain is to produce a large plastic zone at the crack tip and thus induce residual compressive stress as well as blunt the crack, it should not be surprising that stress relief decreases the effectiveness of the prestrain by eliminating residual compressive stress. Moreover, as long as some crack blunting remains, the toughness should still be higher than the base condition value.

These are the results obtained from these experiments. They suggest, at least for the high levels of notch prestrain, that both crack blunting and residual compressive stress do play a role in toughness improvement from warm prestressing. However, thermal stress relief was not helpful in improving toughness after strain aging and, in some cases, was distinctly harmful. This is entirely contrary to the influence of thermal stress relief on the impact test specimens where stress relief was shown to improve toughness.

The stress relief treatments when applied to the uniformly prestrained A516 grade 70 steel produced differing results, as may be seen by comparing specimens B and B_{SR} and C and C_{SR} on Figs. 9 and 10. In this instance, specimen B_{SR} had much lower toughness than specimen C_{SR} , although the strained and aged toughnesses were nearly equivalent. The difference in these two sets of specimens was the sequence of specimen preparation. Specimens B_{SR} were fatigue cracked prior to stress relief while specimens C_{SR} were fatigue cracked after stress relief.

Since these uniform strain specimens were small (0.60 in. thick for A516 grade 70) and the loads em-

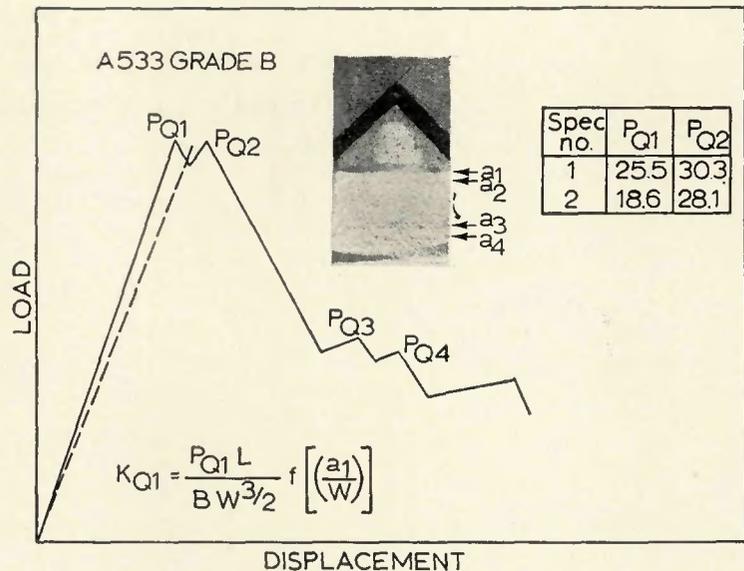


Fig. 16—Load vs. deflection curve and fracture surface for A533 Grade B strained, aged and stress relieved

ployed were difficult to control, the K level during fatigue cracking of these specimens (and only these four specimens) was higher than recommended. It is possible that the fatigue cracking procedure after stress relief raised the K_{Ic} level above that of B_{SR} by inducing plastic strains at the crack tip. However, it may also be suggested that the presence of the fatigue crack itself, with the plastic zone that is always associated with it, is a prior condition for this toughness loss. For the A533B specimens, where fatigue cracking was more carefully controlled, the loss between C and C_{SR} is equally great.

Another interesting feature of this unexpected result was the atypical load-displacement records and fracture surfaces of the uniform prestrain specimens exhibiting these low K_{Ic} numbers after stress relief. All four specimens (2 of A533B and 2 of A516 grade 70) showed load vs. displacement plots and fractured surfaces such as the example in Fig. 16. These results indicate that the crack experienced several cycles of popping through a locally embrittled region and then being arrested.

An effort was made to correlate the most distinct load points (P_{Q1} , P_{Q2} , P_{Q3} , P_{Q4}) with the most distinct crack front advancements on the fractured surfaces (a_1 , a_2 , a_3 , a_4). It was difficult to discern what were and were not crack front advancements on the fractured surfaces; hence, only the first and second most distinct load points were correlated with the first and second most distinct crack fronts respectively. The results of this correlation, as shown on Fig. 16, indicate that the K_{Q2} values were higher than the K_{Q1} values. This suggests that the

crack ran from a region of low toughness into a region of higher toughness and was arrested. This behavior would suggest that the conditions in the region at the crack tip were made more severe by the stress relief treatment.

Conclusions

The results of this experimental investigation may be summarized as follows:

1. Charpy impact and tension test data indicate that strain aging embrittlement occurs in A516 grade 70 steel and A533B steel, but is more severe for the A516 grade 70 than for the A533B.
2. Charpy impact and tension test data indicate that strain aging embrittlement is more severe for the repeated straining and aging condition than for the single cycle strain age condition.
3. The study of different aging temperatures showed that 500 and 650° F produced almost the same response while 400° F produced slightly lesser response in the A516 grade 70 steel and negligible response in the A533B steel.
4. Stress relief treatments following straining and aging treatments improved the Charpy impact toughness of both steels.
5. The fracture toughness data for uniformly prestrain steels showed a moderate decreasing trend in K_{Ic} for specimens that had been strained and aged.
6. The notch strain investigation showed that the benefits of warm prestressing were degraded by subsequent aging treatment.
7. High levels of notch strain at 200° F followed by aging at 650° F

produced K_{Ic} values that were significantly higher than the base plate K_{Ic} values. The K_{Ic} values resulting from the high notch strain followed by an aging treatment were still above the K_{Ic} of the unstrained plate. The low level of notch strain followed by aging produced toughness values below the K_{Ic} level of the plate.

8. Both uniform and notched pre-strain fracture toughness studies at -150° F indicated that thermal stress relief treatments, either after prestraining or following aging, did not increase the toughness of the material tested. In some cases, the fracture toughness of the plate was decreased by this treatment.

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