FRACTURE CHARACTERISTICS OF
METASTABLE AUSTENITIC STEELS UNDER
CYCLIC LOADING

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ABSTRACT

The behavior of a class of metastable austenitic steels, called TRIP steels, under cyclic loading was investigated. The alloy composition was chosen to have the M_s well below room temperature and the M_D above room temperature after thermo-mechanical processing. Both high-strain low cycle experiments on round and well-polished specimens, as well as fatigue crack propagation (fcp) tests on SEN specimens at various stress-intensity range (ΔK) levels were carried out. To study the effect of a mixed austenite-martensite matrix, low cycle fatigue tests were also done on the TRIP steel after inducing fresh martensite by a very high pre-strain. To establish the role played by the martensite transformation, tests were also run at 200°C, which was above M_D . The amount of martensite induced was magnetically measured by a "permeameter" built specifically for this purpose.

It was found that the low cycle fatigue life of TRIP steels both at room temperature (in the presence of martensitic transformation) and at 200°C (in the absence of the transformation) were related to the plastic strain range, ϵ_{PR} , by the Coffin-Manson law. Either cyclic hardening or softening occurred at room temperature depending primarily upon the ϵ_{PR} used in cycling. Hardening was observed for ϵ_{PR} greater than approx. 3%, while softening occurs below 3% ϵ_{PR} . For 200°C low cycle fatigue tests, only cyclic softening was observed in all the cases.

A simple theoretical model of fcp based on fracture mechanics was developed. To a first approximation, the experimental results were in agreement with the model and showed the correlation between the ΔK and fcp rates as $da/dn \propto (\Delta K)^4$. The fatigue fracture appearance of TRIP steels comprised of fatigue striations, quasi-cleavage and elongated dimples reflecting the extremely complex structure of TRIP steels.

The alloy deformed 80% at 250°C showed better fcp properties than a number of alloy steels of similar strength levels and compared favorably with maraging steels in the low ΔK range.

I. INTRODUCTION

After thermomechanical processing, certain metastable austenitic steels have been found to yield unusual combinations of strength, ductility and toughness.¹⁻⁵ This particular class of steels has been designated as "TRIP" steels. It has been shown that the high toughness and ductility associated with these steels are due to the transformation of metastable austenite to martensite during the tensile testing at temperatures below M_D .¹⁻⁶ (M_D is the temperature above which no martensite can be induced by deformation.)

At a test temperature below M_D there is a critical strain or stress beyond which the parent austenite partially transforms to martensite. The transformation to martensite causes strengthening of the deformed region. Because of this, the subsequent deformation is forced into adjacent material. Necking is inhibited and the uniform elongation of austenite is increased. The behavior of these steels under monotonic loadings has been studied extensively as a function of test temperature,^{2,5,7} and amount and temperature of prior deformation.^{1,2,5} However, no work has yet been reported concerning the behavior of these steels under cyclic loading. From an engineering application viewpoint the fatigue property of any material is of significant importance as a majority of service failures are caused by fatigue fracture.

Hence, the present investigation was planned to study the behavior of these steels under cyclic loading. For this purpose, an alloy of such

composition was chosen as to have the M_s temperature well below room temperature and the M_D above room temperature after thermo-mechanical processing.¹ (M_s is the temperature below which martensitic transformation takes place on cooling.)

The steels were austenitic at room temperature. Whenever a certain amount of plastic strain was induced in the material, some austenite transformed to martensite. In the region of the low-cycle fatigue failures, i.e. for lives of 10^4 cycles or less, operative stresses were above the yield stress of the material and hence plastic strain was involved in each cycle. Thus during cyclic loading and unloading, the plastic strain accumulated with each cycle, thereby causing more and more transformation of austenite to martensite. Therefore the low cycle region of failures in TRIP steels offers a challenging area for study.

During the last two decades, an increasing amount of information on low-cycle fatigue behavior of metals has been published. Either load or strain was usually maintained constant in any particular test. In general, the results of constant-load, low cycle fatigue tests are presented in the form of conventional S-N curves where S and N are respectively the nominal stress or stress range and the corresponding cyclic life of the specimens. Although the shape of a typical S-N curve can be qualitatively described, it is difficult to make a precise analysis for this type of test. On the other hand, results of constant strain, low cycle fatigue tests have consistently shown a linear relationship between the plastic strain range and the number of cycles to failure on a log-log basis.⁸⁻⁹

Empirical relationships have been developed relating the plastic strain range to the low cycle fatigue life of metals. The one commonly observed in metals is the Coffin-Manson law:⁸⁻¹³

$$\epsilon_{PR} N^{\alpha} = C \quad (1)$$

where

ϵ_{PR} is the plastic strain range

N is the number of cycles to failure, and α and C are constants

It is desirable to establish the effect of a strain induced martensitic transformation on this law. Thus, room temperature low cycle fatigue tests were carried out at different cyclic strain levels. To study the effect of cyclic loading on a mixed austenite-martensite matrix, tests were also done after very high prestrain (approx. 10-20%). This high value of prestrain induced martensite in the parent austenite thereby giving a mixed austenite martensite matrix. The amount of martensite so induced was measured in a "permeameter" built specially for this purpose. In order to establish the role played by the martensitic transformation on low-cycle fatigue, base-line data were obtained from tests run at 200°C. These represented the properties of austenite since 200°C was above the M_D for this alloy and hence, no transformation took place during cycling.

Another aspect of fatigue which has received considerable attention in the last decade is that of fatigue crack propagation. This aspect is quite important from the practical view-point because the typical "S-N curve" and low cycle fatigue concept determine a "safe-life" only for a smooth and well polished specimen. Hence these approaches do not account for preexisting flaws in the material or accidental damage to

them during production or service. It is also important from the theoretical viewpoint since nominal fatigue lives consist of a crack initiation and a propagation stage. Separation of these two components requires detailed information about the propagation stage. Hence, a fracture mechanics approach is made to analyse the growth of fatigue cracks. This analysis indicates that the crack propagation rate depends approximately on the fourth power of the stress-intensity factor range,¹⁴⁻¹⁷ i.e.

$$da/dn \propto (\Delta K)^4 \quad (2)$$

where

da/dn is the fatigue crack propagation rate, and

K is the stress intensity factor range ($\Delta K = K_{\max} - K_{\min}$)

The stress intensity factor is a parameter which depends upon the applied stress, the crack-length and the geometry of the specimen. $K_{\max}$ and $K_{\min}$ are the maximum and minimum values of stress-intensity-factor applied to the specimen. In order to establish fatigue crack propagation as a function of stress intensity factor range for TRIP steels, single-edge notched (SEN) specimens were utilized.

In addition, the fracture surfaces of both low-cycle fatigue and fatigue crack specimens were observed using

- (i) Scanning electron microscope
- (ii) Transmission electron microscope (replicas of fracture surfaces were observed).

The scanning electron microscope was used because of its large depth of focus and ease of fracture surface observation while the transmission

electron microscope was used because of its high resolution and capability to bring out structural details.¹⁸⁻²⁰ For the present investigation an alloy of nominal composition of 9Cr, 8Ni, 4Mo, 2Mn, 1Si and 0.25 carbon was used. This alloy was given the following four thermo-mechanical treatments before the experimentation

- (i) Deformed 80% at 450°C
- (ii) Deformed 80% at 250°C
- (iii) Deformed 20% at 450°C
- (iv) Deformed 20% at 250°C

For all these treatments, low cycle fatigue tests were run both at room temperature and 200°C. Another series of room temperature low cycle fatigue tests was run after a large amount of prestrain. Fatigue crack propagation tests on SEN specimens were also run at various ΔK levels for all these treatments.

II. EXPERIMENTAL PROCEDURE

A. Material Selection and Preparation

The alloy composition was chosen from previous work with TRIP steel.¹⁻⁵ A suitable balance of various alloying elements and carbon was made so that the metastable austenitic steels had M_s temperatures well below room temperature, while the M_D temperatures were above room temperature after thermomechanical processing. The austenite was designed to have a high work hardening rate, high stability and extensive precipitation hardening with prior deformation at a suitable elevated temperature. The compositions used are shown in Tables I and II.

The steels were prepared by induction melting of high purity elements under vacuum. The ingots so prepared were homogenized at 1100°C for three days. The ingots then went through two different series of thermomechanical processing depending upon whether the final product was meant for low cycle fatigue tests or fatigue crack propagation tests.

For low cycle fatigue testing the homogenized ingots were hot forged at 1100°C to 1-1/4 in. diam. (for a subsequent reduction of 80% in area and 5/8 in. (for a subsequent reduction of 20% in area) round bars. The bar stock was cleaned of surface scale by sand blasting and acid pickling prior to further heat treatment. The bars were austenitized at 1200°C for one to three hours under an atmosphere of 4% hydrogen in helium (forming gas) and then brine quenched. The material was then form-rolled at 250°C and 450°C to 9/16 in. dia. round bar.

For fatigue crack propagation specimens, the homogenized ingots were hot forged at 1100°C to 3/8 in. thickness (for subsequent 80% reduction)

and 1/10 in. thickness (for subsequent 20% reduction) flat pieces. After cleaning the surface by sand blasting and acid pickling, these bars were also austenitized at 1200°C for three hours under an atmosphere of 4% hydrogen in helium (forming gas) and then brine quenched. The material was then flat rolled at 250°C and 450°C to 0.075" thick plates. In order to maintain close temperature control during rolling, the rolls were preheated and the material was reheated in an electric furnace between passes.

B. Mechanical Testing

1. Tensile Testing

Tension tests were carried out on both round (for comparison to low-cycle fatigue) specimens as well as flat (for comparison to fatigue-crack propagation) specimens. The flat tensile specimens were made from the plates fabricated for fatigue crack propagation test specimens. The specimens were machined in such a way that the tensile axis was in the rolling direction of longitudinal direction. Figures 1 and 2 show both the round and flat tensile specimens.

The round specimens were tested on a Materials Testing System(MTS) 300 KIPS capacity universal testing system. These tests were performed at a cross head speed of 0.001 in/sec. at both 25°C (room temperature) and 200°C. For 200°C tests, the specimens were immersed in an oil bath which was maintained at constant temperature by means of a heating element. The flat specimens were tested in a 5,000 kg Instron machine using the same strain rate as that used for the tensile testing of round specimens.

2. Low Cycle Fatigue Testing

Low cycle fatigue specimens as shown in Fig. 1 were machined from the processed round bars. The specimen surface was finely polished

by 600 grit papers in order to avoid any possible surface effects on the low cycle fatigue. All the low cycle fatigue tests were carried out in the MTS 300 KIPS capacity machine under "push-pull" conditions. A specially designed jig as shown in Fig. 3 was used for room temperature testing. This jig gives good alignment and diminishes the slackening during the compressive cycle.

For 200°C tests the specimen was surrounded by hot oil contained in a cylindrical container. The bottom of the container had a flanged hole which fitted around the grips. A leakproof seal was obtained by means of O-rings. Figure 4 shows the sketch of this high temperature fixture. The silicone oil used to achieve the temperature was heated by heating elements. The temperature was controlled by two thermo-couples, one at the specimen surface while the other was near the bottom of the bath. The stability of the bath was such that the temperature difference between the two thermocouples would never exceed 3°C during the entire duration of any fatigue test.

The tests were strain controlled; a diametral clip gage was used to measure the change in diameter. Figure 5 is a photograph of the complete assembly for room temperature testing. The output of the strain gage together with that of the load cell was recorded on an X-Y recorder at appropriate times to give the hysteresis loop. In addition, a high speed strip chart recorder was also employed which continuously recorded the load as well as strain throughout the duration of the fatigue test. All the low-cycle fatigue tests were conducted at a cycling rate of 0.1 cycles per second.

3. Fatigue Crack Propagation Testing

Fatigue crack propagation tests were performed on single-edge-notched, 0.075 inch thick specimens shown in Fig. 6. These specimens were machined from the rolled plates in such a way that the tensile axis was in the rolling direction and the notch was at right angles to the rolling direction. The specimen design accommodated pin loading for good alignment and ease of preparation.

Fracture tests were performed on them at a specimen extension rate of 0.01 in./sec. All the fracture as well as fatigue crack propagation tests were performed on the MTS machine.

The SEN (single edge notched) specimens were tested under tension-tension fatigue at a cyclic speed of four cycles per second. In these tests, load was the controlled variable. Here the load as well as the length of the stroke during cycling was continuously recorded on the high speed chart recorder throughout the duration of the test. An X-Y recorder was also used to monitor the load and stroke. Both fracture as well as fatigue crack propagation specimens were prefatigue cracked at a cycling rate of 4 CPS. The pre-fatigue cracking was necessary in order to obtain a sharp notch which could not be easily obtained by machining.

C. Measurement of the Amount of Martensite

The amount of martensite transformed from the parent austenite was measured magnetically. This was possible because austenite is nonmagnetic (paramagnetic) while martensite is magnetic (ferromagnetic). Two different "permeameters" were used in order to measure the saturation induction of the specimens. One of these was for measuring the saturation

induction of the sheet tensile specimens while the other one was for the low cycle fatigue specimens. Figures 7 and 8 schematically illustrate the two permeameters. In both these permeameters, two detecting coils bucking each other were placed between the poles of an electromagnet. The exciting current in the coils of the electromagnet could be switched from a negative value to a positive value so that the magnetic field between the poles could be reversed.

In the tensile permeameter the signal from the detecting coils was integrated and recorded on a chart recorder. The signal from the two search coils (without any specimen) was balanced by a divider to give a minimum signal. This was done to minimize the imperfectness in the bucking of the two coils. Any increase in the signal obtained with the specimen inserted in one of the search coils was attributed to the additional flux in the specimen. This increment was recorded as the flux change in the specimen as given by

$$\Delta\phi_{\text{spec}} = \Delta(B-H) NA_{\text{spec}} = B_S NA_{\text{spec}} \quad (3)$$

where

- ϕ_{spec} = is the flux in the specimen (Maxwells)
- B = is the induction (Gauss)
- B_S = is the saturation induction (Gauss)
- H = is the magnetic field (Oersteds)
- A_{spec} = is the cross-section area of the specimen (cm^2)
- N = is the number of turns in the search coil

A square loop flux standard was used to calibrate the integrated signal.²¹ The integrated voltage resulting from a known variation in the flux, $\Delta\phi_{\text{SIFS}}$ by the flux standard was read on the recorder, i.e.,

$\Delta\phi_{SLFS}$ gave n_0 volts on the recorder. The saturation induction, B_S , is related to the measured voltage, n , from the integrated signal as follows:

$$B_S = 1/2 \frac{l}{NA_{spec}} \times \Delta\phi_{SLFS} \times \frac{n}{n_0} \quad (4)$$

The right hand side has been divided by a factor of two because of the fact that the magnetic field was switched from a positive value to a negative value. This switching was done to eliminate the error due to the zero of the induction in the specimen. This zero is not easily attainable since the remanent magnetization depends on the nature of each specimen.

The signal due to the additional flux in the specimen was corrected for the imperfect bucking of the search coils. This was done by subtracting the minimum signal obtained without the specimen from the total signal read on the recorder with the specimen.

The permeameter used for tensile specimens was designed in such a way that it measured the martensite content in the middle portion of the specimen as shown in Fig. 7, i.e., here the middle 1/2 inch of the gage length was in the air gap of the electromagnet. The permeameter designed for the determination of martensite in low cycle fatigue test specimens differed from the tensile permeameter only in the structure and method of using it, otherwise the principle was the same for both the permeameters. In the low-cycle fatigue tests, most of the austenite-martensite transformation took place near the specimen surface. Only 1/10" of the specimen from the fracture surface was used for measurement. Hence the specimen could pass through only one pole tip of the permeameter as shown in Fig. 8. The broken end was ground flat to minimize the air gap between the sample end and the mating pole face.

The specimen diameter was approximately 2.5 times the air gap while in the tensile permeameter the specimen thickness was 0.10 times the air gap. This required a modification in measuring technique because of large uncertainty of H in the (B-H) calculation. The modification involved using a calibrated specimen that produced nominally the same signal as the test specimen. The calibrating specimen was a cylindrical steel bar of approximately the same diameter as the test specimen. A set of such specimens of varying thickness was prepared to give different known saturation induction values. The fatigue specimens were bucked against the nearest matching calibrating specimen. The unbalanced signal was recorded in the same manner as in the case of the "tensile permeameter."

Hence, saturation induction, B_S in this case is given by

$$B_S = B_S(\text{steel}) \frac{A_{\text{steel}}}{A_{\text{spec}}} + 1/2 \frac{1}{NA_{\text{spec}}} \times \Delta\phi_{\text{SLFS}} \times n/n_0 \quad (5)$$

where

$B_S(\text{steel})$ is the saturation induction for the calibrating specimen

A_{steel} is the cross-sectional area of the calibrating specimen

The amount of martensite for both the low cycle fatigue and tensile specimens was determined by the B_S measurements, assuming B_S proportional to the amount of magnetic phase which is martensite. Thus, the percent martensite can be calculated as follows

$$\text{percent martensite} = B_S/B_0 \times 100 \quad (6)$$

Here B_0 is the saturation induction of a completely martensitic specimen. This value varies with the alloy content and hence was corrected for the composition used.²²⁻²⁵

D. Metallography and Fractography

1. Metallography

Metallographic samples were cut from the low cycle fatigue specimens as well as from the single edge notched specimens. Samples from the grip area represented the unstrained area, while that from the fracture area represented the fatigue area.

Specimens were prepared by wet grinding in several stages to a finish equivalent to number 600 grit paper. They were then electro-polished in a solution of 90% acetic and 10% perchloric acid at 0°C (20 volts). A solution of 5.0 gm cupric chloride, 100 ml hydrochloric acid, 100 ml ethyl alcohol and 100 ml distilled water was used for etching. Observations were performed on a Carl Zeiss Optical Microscope.

2. Transmission Electron Fractography

A transmission electron microscope was used to study the fracture mechanism in finer detail. This was done by examining a carbon replica of the fracture surface. A "two stage" plastic-carbon technique was used to replicate the fracture surface.²⁶

In this technique, a piece of plastic (cellulose acetate) tape was softened in a solvent (methyl acetate or acetone) and forced down upon the fracture surface in such a way that the softened plastic assumed the shape of the fracture surface. After the plastic completely dried, it was stripped off. This was done a number of times to ensure a clean surface, before any replicas were retained. The replicas were

shadowed with platinum-palladium at an angle of approximately 45° in a vacuum of 10^{-4} Torr. This shadowing by the evaporation of platinum-palladium was done in order to accentuate the features on the plastic. The replicas were then coated with carbon by evaporation. Paraffin was placed on the coated side of the replica to strengthen it during the dissolution of the plastic. The carbon replicas after dissolution were transferred into reagent grade acetone to remove any paraffin and were placed on copper grids. These grids were then observed in a Siemens Elmiskop operating at 80 kV using a double tilt specimen holder.

3. Scanning Electron Fractography

The fracture surfaces were also studied by using a scanning electron microscope at magnifications up to 6000X. The unit used was a Model JSM-2 provided by Japan Electron Optics Company. The button type specimens used in the scanning electron microscope were machined from the fracture surfaces of low cycle fatigue specimens.

Before examining, the fracture surfaces were thoroughly cleaned by placing and stripping off the plastic tape in the same way as was done for transmission electron fractography.

III. THEORETICAL CONSIDERATION

A. Low Cycle Fatigue

Tension-compression fatigue testing was done by controlling the total strain. In these tests, the transverse strain range (diametral) rather than the axial strain range was maintained constant. When analyzing the data, the life was taken as the number of cycles required to cause complete rupture of the specimen. It is important to make a distinction here because some investigators regard the life of a specimen as the point at which cracks begin to appear on the surface.⁸

Strain cycling is usually performed between two constant strain values, the maximum strain, $\epsilon_{\max}$, and the minimum strain $\epsilon_{\min}$. The average value of these is referred to as the mean strain, i.e.,

$$\epsilon_o = 1/2 (\epsilon_{\max} + \epsilon_{\min}) \quad (7)$$

The difference of the two values is referred to as the total strain range, i.e.,

$$\epsilon_{TR} = \epsilon_{\max} - \epsilon_{\min} \quad (8)$$

This type of loading is slightly different from what is done conventionally in fatigue testing, wherein the load is the primary variable that is controlled.

Figure 9 shows a characteristic situation in which a specimen is strained in cyclic manner between a maximum strain, $\epsilon_{\max}$, and a minimum strain, $\epsilon_{\min}$. During the loading part of the first cycle, both the elastic plastic strain contribute to the strain corresponding to $\epsilon_{\max}$. This is considered to be the 1/4-cycle point. In the figure the letter A refers to this part of the cycle. The direction of the load is then reversed. This reverse

loading is continued until the diameter required to produce the predetermined minimum strain, $\epsilon_{\min}$ is obtained. This strain is compressive in this case. Point C corresponds to this stage in the cyclic process ($N = 0.75$). N refers to number of strain (cycles). As can be seen from Fig. 9 a compressive load is required to get the zero strain (point A' in Fig. 9). Cycling then proceeds between these two extreme values of strain, i.e., $\epsilon_{\max}$ and $\epsilon_{\min}$. This soon results in a characteristic hysteresis loop. Further changes in this hysteresis loop occur rather slowly. Figure 10 shows a typical hysteresis loop, the dimensions of which can be described by its width ϵ_{TR} , the total axial strain range, and its height $\Delta\sigma$, stress range. The total axial strain range ϵ_{TR} consists of elastic and plastic components, i.e.,

$$\epsilon_{TR} = \epsilon_{ER} + \epsilon_{PR} \quad (9)$$

where

ϵ_{ER} is the axial elastic strain range

ϵ_{PR} is the axial plastic strain range

The first observation made whenever a specimen is strained in the manner discussed in Fig. 9 is that it requires different loads to accomplish a desired amount of strain, depending on the number of prior applications of the strain cycle. This fact is illustrated in Fig. 11 and Fig. 12. These curves show the stress range as a function of number of applied cycles for two different cases. The curve (a) illustrates the behavior of one class of materials described as cyclic strain hardening materials while the curve (b) shows that for another class of materials known as cyclic strain softening materials. In the former class, the stress range increases with the number of cycles, while in the latter class, the stress

range decreases with the number of cycles. In most cases the stress range reaches a saturation value beyond which no hardening or softening occurs.

In the present tests, the total diametral strain range was controlled. Since the Coffin-Manson Law is for plastic strain range, the part of the total strain range which was elastic had to be separated from that which was plastic. The elastic strain range is simply the stress range divided by the elastic modulus. The plastic range is obtained by subtracting the elastic strain range from the total strain range. In the present investigation, the axial strain range was deduced from a knowledge of stress range, diametral strain range and the elastic constants as follows:

$$\epsilon_t^d = \epsilon_{el}^d + \epsilon_p^d \quad (10)$$

where

ϵ_t^d is the diametral total strain range
 ϵ_{el}^d is the diametral elastic strain range
 ϵ_p^d is the diametral plastic strain range

Hence,

$$\epsilon_p^d = \epsilon_t^d - \epsilon_{el}^d \quad (11)$$

Also, from elasticity relations,

$$\epsilon_{el}^d = \mu \Delta\sigma / E \quad (12)$$

where

μ is Poisson's ratio

$\Delta\sigma$ is the stress range

E is the modulus of elasticity.

Also from elasticity relationships,

$$\epsilon_{El} = \Delta\sigma / E \quad (13)$$

where ϵ_{El} is the axial elastic strain range. Now, in the plastic range, because of constancy of volume ,

$$A_0 l_0 = A l$$

where

A_0 is the original cross sectional area

l_0 is the original gage length

A is the cross-sectional area at any instant (unloaded)

and l is the gage length at that instant. (unloaded)

i.e.,

$$l/l_0 = A_0/A$$

Taking the natural logarithm of both left and right hand sides, we have

$$\ln l/l_0 = \ln A_0/A.$$

Substituting $\pi D_0^2/4$ and $\pi D^2/4$ for A_0 and A respectively in the above expression, we have

$$\ln l/l_0 = \ln \frac{\left(\frac{\pi D_0^2}{4} \right)}{\left(\frac{\pi D^2}{4} \right)} = 2 \ln D_0/D$$

here D_0 is the original diameter of the specimen while D is diameter at any instant (unloaded).

But,

$$\ln D_0/D = \epsilon_p^d$$

and

$$\ln l/l_0 = \epsilon_{PR}$$

where ϵ_{PR} is the axial plastic strain range hence

$$\epsilon_{PR} = 2\epsilon_p^d \quad (14)$$

Solving from Eqs. (11), (12) and (14).

$$\epsilon_{PR} = 2(\epsilon_t^d - \epsilon_l^d)$$

i.e.,

$$\epsilon_{PR} = 2(\epsilon_t^d - \mu \Delta \sigma / E) \quad (15)$$

From Eqs. (9), (13) and (15)

$$\epsilon_{TR} = 2 \left(\epsilon_t^d - \frac{(\mu - 0.5)\Delta\sigma}{E} \right) \quad (16)$$

where ϵ_{TR} is total axial strain range.

In the present investigation the modulus of elasticity and Poisson's ratio were taken to be 30×10^6 psi and 0.3 respectively. The diametral strain range was measured by means of a specially designed resistance strain gauge which could be attached to the specimen as shown in Fig. 5. Since the stress range does vary with the number of cycles, the saturation value of stress range was used in the plastic strain range calculations. For the cases where there was no saturation stress range value, the value of the stress range halfway through the life of the specimen was taken.

B. Fatigue Crack Propagation

Fatigue crack propagation tests are normally conducted under tension tension loading. During these tests either a constant cyclic stroke amplitude or a constant cyclic load amplitude is maintained. In the present investigation the applied cyclic load range was maintained constant throughout the duration of one run.

The crack propagation rate was determined by measuring the increase in crack length, Δa , after cycling for a definite number of cycles, ΔN , so that

$$\frac{da}{dn} = \frac{\Delta a}{\Delta N} \quad (17)$$

The fatigue crack growth rate, da/dn , of a crack for a given material and environment depends upon the nature of the load time history and the configuration of the structure including the crack.

The fracture mechanics approach allows loading and configuration effects to be described in terms of a single parameter, i.e. the stress-intensity factor. The stress intensity factor is a parameter which depends upon the applied stress, the crack length, and the dimensions of the specimen. This parameter was first introduced by Irwin,^{27,28} who derived expressions for stresses in the vicinity of a crack tip, assuming the crack to be contained in a two-dimensional plane sheet of isotropic elastic material. Using the coordinates shown in the figure below:

the stresses at a point P are given as below:²⁹

$$\begin{aligned}\sigma_x &= \frac{K}{\sqrt{(2\pi r)}} \cos \theta/2 (1 - \sin \theta/2 \sin 3\theta/2) \\ \sigma_y &= \frac{K}{\sqrt{(2\pi r)}} \cos \theta/2 (1 + \sin \theta/2 \sin 3\theta/2) \\ \tau_{xy} &= \frac{K}{\sqrt{(2\pi r)}} \sin \theta/2 \cos \theta/2 \cos 3\theta/2\end{aligned} \quad (18)$$

where

K is the stress-intensity factor

2a is the crack length

r and θ are the polar co-ordinates of the point P

σ_x is the normal stress component in the x (small x) direction

σ_y is the normal stress component in the y direction

τ_{xy} is the shear stress component on the plane perpendicular to the x direction, and acting in the y direction.

In general, the stress fields near crack tips can be divided into three basic types,^{30,31} each associated with a local mode of deformation of the crack surface, as illustrated in Fig. 12. The three modes are:

Mode I. The opening or tensile mode: This is associated with the local displacement in which the crack surfaces move directly apart. Here tensile force is applied normal to the face of the crack.

Mode II. Edge sliding or shear mode: In this the crack surfaces slide over one another normal to the leading edge of the crack.

Mode III: Tearing Mode: Here the crack surfaces slide with respect to one another in a direction parallel to the leading edge.

The superposition of these three modes is sufficient to describe the most general case of crack-tip deformation and stress fields.

Since the stress-intensity factor describes the effect of both external loading and configuration on the stress field surrounding the growing crack tip, the rate of fatigue crack growth should depend on the stress-intensity-factor. This hypothesis has been investigated by various workers, and they have found a correlation between the stress-intensity factor and crack propagation rate.^{14-17, 32-33} The empirical relationship obtained by several investigators is

$$da/dn \propto (\Delta K)^m$$

where

ΔK is the stress intensity factor range

m is a numerical constant.

However, from the Irwin expressions (Eq. 18) for stress at the crack tip, it can be seen that the stress approaches infinity at the crack tip. Actually, before the stress can reach this value, the material just ahead of the tip plastically deforms and as a result, there is a plastic zone just ahead of the crack tip. Hence in the present analysis the crack tip plastic zone has been included so as not to invalidate the approach.

In order to establish a quantitative relationship between fatigue crack growth rate and stress intensity factor, it has been assumed that the growth is governed by the accumulation of damage due to cyclic plastic strain. In the past this hypothesis has been used by several investigators.^{14-16, 34-36}

Therefore in order to establish the fatigue crack growth rate, the following are needed:

(1) A quantitative fracture criterion that permits its application to the case of cyclic loading, i.e., how much damage will have to take place before the crack propagates.

(2) The elastic-plastic distribution of stress and strain in front of the growing crack which takes the plastic zone into account. As yet no convenient analysis of the elastic-plastic stress and strain distribution in front of a crack under tension is available. However since this problem has been solved for the longitudinal shear mode (Mode III) by McClintock and others³⁷⁻³⁹ the tensile mode (Mode I) analog from the Mode III elastic plastic solution will be used.

Various fracture criteria have been used in the past.^{31,39} In the present investigation a simple and realistic criterion based on actual low-cycle fatigue experimental observations was used. It will be shown later that for constant strain amplitude tests on "TRIP" steels, the following law holds

$$N_f^{1/2} \epsilon_{PR} = \epsilon_f / 2 \quad (19)$$

where

N_f = cyclic number of cycles to failure

ϵ_{PR} = cyclic plastic-strain range

ϵ_f = fracture strain under monotonic loading

From the expression (19) it follows:

$$2N_f^{1/2} \times \epsilon_{PR} / \epsilon_f = 1$$

i.e.

$$4 N_f [\epsilon_{PR} / \epsilon_f]^2 = 1 \quad (20)$$

From this, a general expression for fracture criterion under fatigue for TRIP steels could be obtained. The expression can be written as

$$\int_0^{N_f} 4 [\epsilon_p^a / \epsilon_f]^2 dn = 1 \quad (21)$$

where

ϵ_p^a is average plastic strain over the region in which the fracture criterion is to be satisfied.

The tensile analogy obtained from Hult and McClintock's strain distribution for longitudinal shear has been shown to adequately describe elastic-plastic strain distributions.^{36,40-41} From this analogy the strain at a point P in TRIP steel can be written as follows:

$$\epsilon = \sigma_{ys} / E (R_p / r) \quad (22)$$

where

σ_{ys} is the yield stress of the austenite-martensite mixture at the onset of plastic flow.

E is the modulus of elasticity

R_p is the distance of the elastic-plastic boundary from the crack tip through the point at which the strain is measured, i.e., the plastic zone size.

r is the distance of the point P from the crack tip.

From this expression, it can be seen that in the elastic-plastic case strain varies inversely with the distance of the point from the crack-tip while in the elastic case (Irwin's expression, Eq. 18) the stress and hence the strain varies inversely with the square root of the distance.

To obtain an expression of fatigue crack propagation for the "TRIP" steel, from the fracture criterion (expression, 21) ϵ_p is needed. This is determined by considering a small region just ahead of the crack tip as shown below.

ϵ_p^a over this region can be obtained by integrating the strain over the region 'l' shown in the Fig. i.e.,

$$\epsilon_p^a = \frac{1}{l^2 \Delta\theta / 2} \int_0^l \epsilon_p \pi \Delta\theta dr \quad (23)$$

All the terms except ϵ_p in this expression are very clearly illustrated in the above figure. ϵ_p is plastic strain at r and can be obtained by subtracting elastic strain from the expression for total strain.

$$\epsilon_p = \frac{\sigma_{ys}}{E} \left[\frac{R_p}{r} - 1 \right] \quad (24)$$

Substituting 22 and 23 in the fracture criterion expression, and solving

$$\frac{da}{dn} = J (\epsilon_y/\epsilon_f)^2 \cdot \frac{R_p^2}{l} \quad (25)$$

where

J is a constant

ϵ_y is the yield strain in tension

According to Irwin⁴², the plastic zone size for loading is given as :

$$R_p^* = \frac{K^2}{\pi \sigma_{ys}^2} \quad (25)$$

On the basis of this, the plastic zone size for repeated loading is given by (see Appendix A)

$$R_p = \frac{(\Delta K)^2}{4\pi \sigma_{ys}^2} \quad (26)$$

Substituting this value of R_p in the fatigue crack rate expression and rearranging, one obtains

$$\frac{da}{dn} = J (\epsilon_y/\epsilon_f)^2 \left(\frac{K^2}{4\pi \sigma_{ys}^2} \right)^2 \frac{1}{l} \quad (27)$$

In the above expression J , ϵ_y , ϵ_f , σ_{ys} and l are constants for a given material and specimen geometry. Hence, the above expression can also be written as:

$$\frac{da}{dn} \propto (\Delta K)^4$$

i.e., the fatigue crack propagation rate depends on the fourth power of the applied stress-intensity factor range.

Experimental results agreeing with this relationship have been reported by Rice¹⁶ and Paris¹⁴⁻¹⁵ for high strength aluminum alloys and high strength steels. In the above analysis, Irwin's oversimplified model for plastic zone size has been used. However, for TRIP steels it has been shown by Gerberich³ et al. that Dugdale's plastic zone model gives an excellent agreement with the effects of strain-induced martensite transformation on crack propagation during fracture toughness testing.

According to Dugdale⁴³ the plastic zone size under unidirectional loading is given as

$$R_p^* = a [\sec (\pi\sigma/2\sigma_{ys}) - 1] \quad (28)$$

where σ is the applied stress.

Hence for cyclic loading, based on the earlier discussion, we have

$$R_p = a [\sec (\frac{\pi\Delta\sigma}{4\sigma_{ys}}) - 1] \quad (29)$$

Substituting, this in the fatigue crack rate expression (25) we have

$$\frac{da}{dn} = \frac{J}{l} \left(\frac{\epsilon_y}{\epsilon_f} \right)^2 a^2 [\sec \frac{\pi\Delta\sigma}{4\sigma_{ys}} - 1]^2 \quad (30)$$

Dugdale's model is for a plate of infinite size. Utilizing Westergaard's finite width correction,³⁰ we have for a plate of width W

$$\frac{da}{dn} = J \left(\frac{\epsilon_y}{\epsilon_f} \right)^2 \left(\frac{\epsilon_y}{\epsilon_f} \right)^2 \left(\frac{W^2}{\pi^2} \tan^2 (\pi a/W) \right) [\sec \frac{\pi\Delta\sigma}{\sigma_{ys}} - 1]^2 \quad (31)$$

The fact that this expression fits the experimental results on TRIP steels can be shown by expanding and simplifying the above expression by neglecting the higher order terms. The final result after simplification can be written as

$$\frac{da}{dn} = A^2 a^2 \left(\frac{\Delta\sigma}{\sigma_{ys}} \right)^4 + 2AB a^3 \left(\frac{\Delta\sigma}{\sigma_{ys}} \right)^6 + B^2 a^4 \left(\frac{\Delta\sigma}{\sigma_{ys}} \right)^8 \quad (32)$$

where

$$A = f(a, w)$$

$$B = f(a, w)$$

For a single edge notched specimen (used in the present investigation),^{44,45}

$$\Delta K = Y \Delta\sigma (a)^{1/2} \quad (33)$$

where

$$Y = f(a/w)$$

Substituting this in fatigue crack expression, we have

$$\frac{da}{dn} = C^2 \left(\frac{\Delta K}{\sigma_{ys}} \right)^4 + 2CD \left(\frac{\Delta K}{\sigma_{ys}} \right)^6 + D^2 \left(\frac{\Delta K}{\sigma_{ys}} \right)^8 \quad (34)$$

$$C = A/Y^2$$

$$D = B/Y^4$$

Thus, we can see while da/dn is dependent on $(\Delta K/\sigma_0)^4$ it may also depend upon the higher powers of ΔK . At lower values of ΔK , of course, the contribution due to the higher powers of ΔK is negligible. But at higher values of ΔK , the higher powers become important.

As discussed earlier in this section, the stress-intensity factor determination requires an expression connecting K with the specimen dimensions (including the crack) and applied load for a particular design

of specimen. Such expressions are determined either by boundary collocation procedures or by stress analysis of the specimen.

During the present investigation the stress intensity values were obtained from the boundary collocation results by Gross⁴⁴ et al. Brown and Srawley⁴⁵ have discussed the merits of this method over other methods. The K calibration is represented by the following equation

$$Y = \frac{KtW}{P a^{1/2}} = 1.99 - 0.41 (a/w) + 18.70 (a/w)^2 - 38.48 (a/w)^3 + 53.85 (a/w)^4 \quad (35)$$

Where t is the thickness of the specimen. A curve representing this relationship is shown in Fig. 13. For calculating stress intensity range, ΔK , $\Delta P (P_{\max} - P_{\min})$ was substituted in the expression (35) in place of P . The rate of a propagation for a particular run was determined as follows

$$\frac{da}{dn} = \frac{a_{\text{end}} - a_{\text{start}}}{\Delta N}$$

where

ΔN was the number of applied cycles

a_{start} and a_{end} were the crack lengths before and after the end of the run.

IV. RESULTS

A. Uniaxial Tensile Test Results

Tables III and IV list the tensile properties for round specimens at room temperature and 200°C respectively. These round specimens were the same as those used for low cycle fatigue studies. Room temperature tensile properties of flat specimens are listed in Tables V. These flat specimens were machined from the same sheet as was utilized for fatigue crack propagation specimens.

In some of the cases (e.g. Alloy D, 80% deformed at 450°C and tested at room temperature) a yield point was observed during tensile testing while in some others (e.g. Alloy D, 80% deformed at 450°C and tested at 200°C) no yield point was observed during tensile testing. In those cases, where a yield point occurred, the yield stress was taken at the point of load drop, i.e., the upper yield point, while in cases where the upper and lower yield points were not clearly defined, the 0.2% offset stress was used. True stress-true-strain calculations were based on measurements of engineering stress-strain data taken from the Instron or M.T.S. recorder (the total elongation as measured at the end of the test was used as the scaling factor.) The elastic strain of both the specimen and the testing machine was subtracted from the total strain in computing these curves. These curves were plotted only up to the point of maximum load on load extension curve. Figures 14 and 15 show the room temperature and 200°C true stress-true strain curves for the four treatments used in low cycle fatigue investigation. In the case of flat specimens, the martensitic transformation was followed during tension testing by a permeameter discussed in Section II. The amount of transformation as a function of percent strain for four representative cases as shown in

Figs. 16 and 17. These figures also show the (broken lines) engineering stress-strain behavior for these cases (full lines).

B. Metallography

Figures 18 and 19 show typical microstructures of the alloy used in present investigation after various thermo-mechanical processing. All the alloys in Fig. 18 have undergone 80% deformation while those in Fig. 19 have undergone 20% reduction in area. Deformation markings in the form of slip traces, deformation twins, and elongated grains can be seen in these figures. All the structures are essentially austenitic. The effect of the increased amount of deformation is apparent in these figures. The austenitic grains are heavily aligned in the rolling direction in the 80% deformed alloys. The rolling direction is indicated by arrows in these figures.

The higher density of deformation markings and other structural features (like carbides) in 18a, and c as compared to 18(b) and (d) can be attributed to the difference in deformation temperatures for the two sets. The alloys in 18(a) and (c) were deformed 80% at 450°C while the alloys in 18(b) and (d) were deformed 80% at 250°C. Furthermore, the microstructures in (a) and (b) are after flat rolling while (c) and (d) are after form rolling. The structures of both the form rolled and flat rolled alloys are essentially similar, the only difference noted is the austenite grain size which is larger in flat rolled specimens.

C. Low Cycle Fatigue Test Results

Three different series of low cycle fatigue tests were run on "TRIP" steels in the present investigation. The objective of the first series

was to determine the number of cycles to failure at room temperature for various cyclic strain levels. Tables VI-IX summarize the data for these tests. The second series was run at 200°C at various strain levels in order to evaluate the fatigue behavior in the absence of the martensitic transformation. The third series was given with a high value of prestrain. Tables XIV - XVIII present the data for the second series while tables X - XIII list the data for the third series. The basis for choosing the stable values of Δ , ϵ_{PR} and ϵ_{TR} has already been discussed in Section III. The reported lives of the specimens are the number of strain cycles required for total separation. Unless otherwise noted, the strain range in the present work represents cyclic axial strain range.

1. Hysteresis Loop Behavior

Hysteresis loops of load vs diameter changes were noted on an x-y recorder from time to time during the tests. Typical loops for one alloy are reproduced in Fig. 20, obtained at room temperature with diametral strain ranges, ϵ_t^d of 8.00 and 2.50.

As will be seen later the considerable degree of cyclic hardening exhibited by loops is largely in Fig. 20a due to the martensitic transformation during cycling. However, for the loops in the Figs. 20b there is no significant transformation to offset the softening of the work-hardened austenite during cycling. Note that for each loop the peak compressive load is greater than the subsequent tensile load.

2. Plastic Strain Behavior with no Prestrain

The results at both room temperature and 200°C for the four TRIP steels investigated are shown in Fig. 21 through Fig. 24 by the log-log plots of plastic strain range against the number of cycles to failure.

One of the purposes of this study was to determine, for the range of diametral strains used, how well the simple Coffin-Manson relationship

$$\epsilon_{PR} N^{\alpha} = C \quad (1)$$

is obeyed. Coffin has suggested $\alpha = 1/2$ as the representative of most material. The exponent $1/2$ means that on a log-log plot, a slope of $-1/2$ should be obtained. In these figures and in fact in all the subsequent figures involving plastic strain range and cycles to failure, straight lines having slopes of $-1/2$ are drawn through the various test points. Furthermore in all the figures the fracture ductility in simple tension, ϵ_f , is represented as ϵ_{PR} at $N_f = 0.25$ cycles. Here, the fracture ductility ϵ_f , is defined as

$$\epsilon_f = \log e (A_o/A_f) \quad (36)$$

where

A_o is the initial specimen area and

A_f is the specimen area after fracture.

The validity of Eq. (1) is measured by the agreement between test points and the straight line having a slope of $-1/2$.

The two test temperatures selected in the present investigation were of special significance. One of them (room temperature) was below the M_D temperature while the other one (200°C) was above M_D . As a result, for 25°C tests, there was a distinct possibility of martensitic transformation taking place during cycling, while no such transformation took place at 200°C . From Figs. 21-24, it is clear that the material had better strain cycling properties at 200°C than that at 25°C . The room temperature LCF* curve extrapolated to the tensile ductility for all the cases except for Henceforth the log-log representation of plastic strain range against cycles to failure will be referred as an LCF curve).

the alloy deformed 80% at 450°C. The LCF curve for this alloy extrapolated to a value somewhat above the tensile fracture ductility. However, for 200°C tests, all LCF curves extrapolated to a value above the tensile fracture ductility except for the alloy deformed 80% at 450°C.

3. Plastic Strain Cycling Behavior With Prestrain

Figures 25-28 show the room temperature LCF curves for TRIP steels after a fairly large amount of plastic strain. In each case the prestrain was more than one-third of the tensile elongation of the material. This amount of prestrain was sufficient to induce a significant amount of martensite. For comparison sake, the LCF curves without prestrain are also shown in these figures. For the LCF curves with prestrain the plastic strain range at $N = 0.25$ has been taken as $\epsilon_f - \epsilon_o$ where ϵ_o is the amount of prestrain. One can clearly see that the curves extrapolate fairly well; the alloy deformed 80% at 450°C being the sole exception which is not quite unexpected as even without prestrain this was the only one alloy which did not extrapolate to the tensile fracture ductility. The value of $\alpha = 1/2$ seems to hold reasonably well, even for the alloys with prestrain. Hence the following modification of Coffin's law yields a good approximation to the experimental data on TRIP steel

$$N_f^{1/2} \epsilon_{PR} = 1/2 (\epsilon_f - \epsilon_o) \quad (37)$$

4. Cyclic Hardening or Softening

Figures 29 through Fig. 32 show the cyclic strain characteristics of TRIP steels at room temperature. Examination of these figures shows that either strain hardening or strain softening can occur for the alloys

investigated, depending primarily upon the amplitude of the strain range used in cycling. The processing of the alloys also has some effect on the strain-cycling behavior. In these figures, the number in the brackets is the plastic strain range used in cycling while the outer number is the total strain range amplitude. From Figs. 29 through Fig. 32 it can be seen that whenever the plastic strain range was greater than approximately 3% in amplitude, there was hardening, while for that below 3% there was softening.

From the strain cycling characteristic plotted for 200°C tests, it can be observed that only cyclic softening took place. Figure 33-36 show the 200°C test behavior for all the four alloys. In these cases a hand magnet showed the fractured specimens to be non-magnetic, thereby indicating the absence of the austenite martensite transformation.

5. Martensitic Transformation in Fracture Area

The percent martensite within 0.1 inch of the fracture surface area is shown in Fig. 37 as a function of the cyclic plastic strain range causing the transformation of the austenite to martensite. As can be seen from the Figure for 200°C cycling, practically no martensite was detected in the fracture area and also, as expected, the amount of martensite formed increased with the cyclic plastic strain range. Another point to be noted is an almost two-fold increase in the amount of martensite for an increase of the cyclic plastic strain range from two to three percent. Further, 80% deformation at 450°C shows a larger amount of martensite per unit plastic strain range than others, thereby indicating that this treatment gives the most unstable structure out of the four

thermo-mechanical treatments investigated. The number in brackets at various points in the figure denotes the cycles to failure for the test represented by the data point. The number of cycles to failure is shown only for room temperature tests, as for 200°C tests, negligible amount of martensite was formed, irrespective of the number of cycles to failure.

6. Observation of Fracture Surfaces

a. Macroscopic Observations: Figure 38 shows the room temperature fatigue fracture surfaces for one of the four processing treatments employed. Two types of fatigue fractures were observed for these steels. A transition range existed in the number of cycles required to produce a given type of fracture. One type of fracture observed for all failures occurring at a lower range of the number of cycles to fracture, was somewhat similar to the cup and cone fracture produced by static tension testing. A shear lip was observed and a reduction in area occurred. For example Fig. 38a shows the fracture surface of a specimen which failed after 25 cycles. It shows a shiny granular central position surrounded by a narrow shear lip. The other type, observed for failures occurring at higher number and cycles, approached a more nearly typical fatigue fracture in which the origin appeared at the surface. This origin was often a parabolic shaped region at the surface where no shear lip existed and little reduction in area was observed. Figure 39 shows the typical fractures schematically.

b. Microscopic Observations: Figures 40-44 show some of the results obtained by the observations of replicas of fracture surfaces by transmission electron microscopy. Figures 40(a) and 40(b) are the fractographs from the center and the shear lip area of tensile fracture surface of the

alloy deformed 80% at 450°C. As can be seen, this material exhibited a wide variety of dimple sizes. The central portion exhibited relatively small and uniform equiaxial dimples while much larger shear type dimples were present on the shear lip. Uniformly elongated shear dimples were observed in the fracture surface of the same alloy after failure at 158 cycles (Fig. 41). Figures 42(b) through 42(d) show tire marks which are frequently observed in low cycle fatigue fracture of high strength materials.²⁶ These fractographs were obtained from the fracture surface of the specimen (80% deformed at 450°C) which failed after 381 cycles. Two distinct modes of failures can be seen in Fig. 43, taken from the replica of the fracture surface of the specimen which failed after 873 cycles. This specimen was also machined from the alloy deformed 80% at 450°C. The top half looks somewhat like striations with superimposed dimples while the bottom half is a rubbed portion which often occurs into reversed cyclic loading conditions. Figure 44 shows fractographs from the fracture surface of the specimen which failed after 365 cycles at 200°C. This specimen was fabricated from an alloy deformed 20% at 450°C. These fractographs do not have well defined features that give the impression that considerable stretching has occurred during the plastic flow process. Figure 44a the deeper indentations may be representative of striation markings.

The fracture surfaces were also examined by using a scanning electron microscope. In the case of low cycle fatigue fracture observations, this was particularly desirable as the fracture surfaces were difficult to replicate because of their geometry and roughness. Figures 45 through 47 show scanning electron micrographs of the fracture surfaces for the

alloys deformed 80% at 450°C and cycled at different strain ranges to failure. Figure 45 was obtained from the fracture surface of a specimen which failed at rather low value of cycles ($N_f=22$). The fracture resembles tensile rupture by microvoid coalescence. Here a wide variation in dimple size was also observed (Figs. 45 (b) through 45 (d)). In Fig. 46(a) the fracture looks laminated which is not surprising considering that the material was heavily rolled to give an 80% reduction in area. Several microcracks can be seen in Figs. 45(b). In low cycle fatigue usually several microcracks are initiated and one of them propagates to failure. These microcracks were also seen on the fracture surface of several other low cycle fatigue specimens. A structure resembling striations is seen in Figs 46(c) and 46(d). Since the number of cycles to failure was only 98 cycles, it is doubtful that these are striations.

Figures 47(c), (e) and (f) show fairly well developed striations. This specimen failed after 1514 cycles and as a result striations are not quite unexpected. Striations are in two packets possibly divided by a subgrain boundary. Another area of fracture surface shows dimples and a notch like groove (Figs 47 and 47b).

Figures 48-51 show scanning electron micrographs of the fracture surfaces for the Alloy A deformed 20% at 450°C and cycled at various strain range levels to failure. This alloy fails predominantly in a ductile manner as can be seen by the scanning electron micrographs from fracture surfaces of specimens failing at as low as 32 cycles (Fig. 48) to as high as 1356 cycles (Fig. 51). In Fig. 49b ($N_f = 64$ cycles) and Fig. 50b ($N_f = 124$ cycles), two regions of dimples separated by a boundary can be seen.

D. Fatigue Crack Propagation Results

1. Crack Propagation Behavior

The crack propagation data for the TRIP steels tested are given in Tables XVIII through XXI. All these were tension-tension tests on single edge notched (SEN) specimens. In order to maintain the accuracy of the results (e.g. calibration of stress intensity factor as discussed in Section III), the maximum length of the crack for which the data was recorded was usually kept below such a value as not to exceed $a/w = 0.65$. Run No. 1 was not included in the Tables XVIII through XXI as during this run a sharp notch was introduced at the tip of the machined crack by fatiguing at low values of load range.

Both the minimum and the maximum load during each run were noted. Usually both the maximum and minimum load remained constant throughout the duration of the test. But for a few runs in the higher stress intensity range, there was a small but gradual change in the maximum value of load. In these cases the values of the load at the beginning and at the end of the run were noted.

The cracklength before and after every run was measured by means of a traveling microscope. Since the crack length changes during a run, the value of ΔK does not remain constant throughout the durations of any one run. Hence the values of ΔK both at the start and at the end of each run are noted. During the present investigations the average of these two ΔK values was taken as the stress-intensity range giving rise to the da/dn during that particular run.

Figures 52-55 present plots of crack-tip stress-intensity range vs growth rate on a log-log basis for the four thermomechanical materials investigated. Through the data points a line of slope of $1/4$ has been drawn. Although there is scatter in the data, it is clear that a relationship of the type $da/dn \propto (\Delta K)^4$ as discussed in Section III provides a very good approximation over the ranges of data for the four TRIP steel.

Although most of the tests were run at 4 cps, some were also run at slower cyclic speed (2 or 1 cps). The slower speed was used for high ΔK 's (i.e. high crack propagation rates). Figure 56 shows the general trend for the four materials. It seems that three of the four TRIP steels show approximately the same crack growth behavior while the fourth one (i.e., 20% deformed at 250°C alloy) shows somewhat higher crack-growth rate than the others for the same stress-intensity range. Among the three alloys showing similar behavior, the 80% deformed at 250°C has slightly better crack growth resistance than the others.

The ratio of P_{max}/P_{min} (.) was usually maintained between 6 and 10. As a result the effect of . if any was negligible.

2. Observation of Crack

The crack path was observed during the test by means of optical (telescopic) equipment. Figure 57 shows the crack path and crack tip for the alloy deformed 80% at 450°C, while Fig. 58 shows the same for the alloy deformed 20% at 450°C. The plastic zone surrounding the crack is very distinctly outlined in Fig. 57(b). It can be noted from the figures that the crack path is straight for 80% deformed alloy as compared to the one deformed 20%. The tensile axis is vertical in these figures.

3. Fracture Surface Observations

Typical fractographs obtained from the replicas of the fracture surfaces of TRIP steel SEN specimens are shown in Figs. 59-62. Figures 59 and 60 are from the fracture surfaces of the alloy deformed 20% at 450°C. Both Figs. 59(a) and (b) show striation marks while Fig. 60 shows tire marks and cleavage features. The striations in Figs. 59(b) are very widely spaced as compared to those in Fig. 59(a). Figures 61 and 62 have been obtained from fracture surfaces of the alloy deformed 80% at 450°C. The fracture surface of this alloy was composed of four different fractographic feature areas which contained well defined fatigue striations (Fig. 62a and b), areas which contain striations not so well defined (Fig. 62a), areas which consist of elongated dimples (Fig. 61c) and areas which showed flat fracture (Fig. 62b). The fractographs shown in Figs. 61(a)-(c) were taken from the fracture surface of the same specimen but from different replicas (specimen No. S-12 in Table XXI).

V. DISCUSSION

A. Tensile Results

The tensile properties of TRIP steels have been discussed in detail by several investigators.^{1,2,5-7} In this section the general features of the tensile results obtained on the alloys investigated are briefly discussed. As seen from Fig. 14 through 17 these steels exhibited a high value of room temperature elongation together with a high value of strength. The enhanced ductility of TRIP steels is due to the austenite to martensite transformation that occurs during testing. The formation of martensite during straining increases the work hardening rates of these steels, and necking of tensile specimens is thereby inhibited. This gives rise to a high value of uniform elongation at room temperature. However, as can be seen in Figs. 14 and 15 there was a sharp drop in elongation at 200°C. Since 200°C was above the M_D temperature of the alloys used in the present investigation, the austenite did not transform to martensite during straining; and the elongation approached that of highly cold-worked austenite.

Alloys with 80% deformation exhibited a yield point and the Luders strain was rather large. Another characteristic observed in the room temperature tensile tests of these alloys was that failures occurred at a strain almost immediately after that corresponding to the ultimate tensile strength, that is, almost all strain was uniform. However, in tests at 200°C considerable necking was observed as indicated by a large percent reduction in area as compared to percent elongation as shown in Table IV.

Both the final amount of martensite and the volume percent of martensite per unit strain observed in a room temperature tensile test were more

for those specimens with 80% prior deformation than those with 20% prior deformation. This indicates that 80% prior deformation makes the alloys more unstable with respect to the strain induced $\gamma \rightarrow \alpha'$ transformation. The reason for this has been attributed to the greater deformation at the elevated temperature causing more alloy carbides to form thereby draining the matrix of solute which raises the M_D . This observation is of special significance in low cycle fatigue results as will be discussed in the next section.

The occurrence of a yield point for alloys deformed 80% may be attributed to the stress-induced transformation; i.e., in these alloys plastic deformation was initiated by the stress-induced formation of martensite.^{4,5} It has been reported by Hall⁶ and Fahr⁵ that the occurrence of Luders strain and yield point indicates that the stress required to initiate the martensitic transformation is less than the flow stress of austenite. Apparently at 20% deformation, the austenite is not strong enough and its flow stress is less than the stress required to initiate the stress induced martensitic transformation. As a result, the shape of the stress strain curve was similar to that obtained for austenitic steels without cold work. This explanation is further substantiated by the slower rate of martensite formation with strain for the 20% deformed alloys as compared to the 80% deformed one as mentioned above.

B. Low Cycle Fatigue Properties

1. Cyclic Hardening or Softening

One of the important aspects of the behavior of TRIP steels is their fatigue resistance during the cyclic hardening. Before proceeding to the discussion of fatigue resistance of TRIP steels, it will be helpful to discuss the phenomenon of cyclic dependent hardening or softening noted in all tests. As mentioned earlier, either hardening or softening can occur under cyclic conditions depending on the initial state of the material and the test conditions. This is reflected by a cycle-dependent change in a host of bulk material properties including hardness and the mechanical hysteresis loop.^{8,10,46}

For each material there is a range of potential strength which can be achieved by various thermal and/or thermal mechanical treatments. It has been postulated that if a material is initially on the low end of this potential strength level, it will cyclically harden while if it is on the high end it will soften.^{47,48} Some intermediate state will be approached which represents a stable condition for the particular material under imposed conditions of cycling.

The initial cyclic rate of change in properties is greatly influenced by the cyclic strain range but rapidly diminishes with repeated cycling. The major changes occur in the first few percent of the fatigue life. As long as the control conditions are not altered, the material quickly adjusts to a nearly stable steady state condition which is reflected by a stable hysteresis loop. By use of the stable values of stress range from several companion tests at different strain ranges a smooth

curve is formed which is called the cyclic stress-strain curve.

For materials which cyclically harden, the cyclic stress-strain curve will be above the monotonic stress-strain curve while for materials which soften the curve is below the monotonic curve.

While the preceding concepts of the softening and hardening of materials with varying degrees of initial cold work are reasonably well established, there is little information in the literature on the cycle dependent softening and hardening of metals which have undergone thermo-mechanical processing, which precipitate harden, and/or transform from one phase to another (e.g. from austenite to martensite) such as is the case for the TRIP steels reported here.

Cyclic stress-strain curves for the TRIP steels investigated at room temperature and at 200°C are shown in Fig. 63 through Fig. 66. The stable values of stress and strain were taken in these figures. Monotonic curves are also shown for comparison. It should be noted that the cyclic stress-strain curves in all cases showed higher strain-hardening rates than the monotonic curves indicating cyclic hardening. For 200°C tests, the strain hardening rate was much lower and somewhat similar to the monotonic behavior. The cyclic stress-strain curve was always below the monotonic stress-strain curve because here the strengthened TRIP steel (because of strain-hardening and precipitation of carbides during prior processing,²⁷) softens during strain cycling. It appears that the hardening effects produced by processing were such that they could be relaxed by strain cycling at 200°C. Of course, this does not mean that the material can be restored to its initial state but rather to some equilibrium condition. Failure by fatigue may occur prior to this. Thus cyclic straining provides a

mechanism for relaxation of those effects which cause dislocation pile ups, precipitate hardening and strain hardening in mechanical or thermomechanical treatments.

Another point noted earlier in stress range vs strain cycle curves for room temperature tests was that there was cyclic hardening above approximately 3% plastic strain range.

Below 3% plastic strain range, there was no hardening but softening (Figs. 29-32). Apparently 3% plastic strain range induced a significant amount of martensite during cycling, while less than 3% plastic strain range did not cause enough transformation to offset the relaxation of some of the causes of strengthening in TRIP steel taking place during cycling. This reasoning could be substantiated by the final amount of martensite in the fractured area of the specimen. As seen in Fig. 37, by cycling beyond 3% plastic strain range, there was approximately a two-fold increase in the amount of martensite after fracture over the amount of martensite induced below 3% plastic strain range. Furthermore, for 200°C tests, where there was no transformation, all the alloys showed softening (Figs. 33 through 36). For these tests, in the absence of transformation, softening was expected since the alloys which are towards the higher end of their potential strength soften and go towards an intermediate value. The softening was much more pronounced in the case of 80% deformed alloys than in the case of 20% deformed alloys. This behavior was not unexpected as 80% deformed alloys are at a much higher end of the potential strength level than that for 20% deformed alloys.

The difference in the hysteresis loop behavior of alloy D (80% deformed at 450°C) cycled at $\epsilon_d^t = 8.00$ and that of the same alloy cycled at $\epsilon_d^t = 2.50$ as noted in Fig. 20, also could be attributed to the martensitic transformation. In the case of $\epsilon_d^t = 8.00$, the strengthening associated with the martensite formed more than offset the tendency of the material to go towards the stable range while for $\epsilon_d^t = 2.50$, the amount of transformation was not enough to offset the softening taking place. The loops in general showed a greater strain hardening in loading from tension into compression than from compression into tension. This could be either due to the fact that the tensile load helps the martensitic transformation while the compressive stress does not⁴⁹, or due to Bauschinger effect, or due to a combination of both effects. The difference in the peak compressive load and the peak tensile load is largely accounted for by the area difference between tension and compression, e.g., for the alloy deformed 80% at 450°C the peak compressive load after 5 cycles at $\epsilon_d^t = 8.00$ was approximately 12,350 lbs., while the peak tensile load was approximately 11,500 lbs. (Fig. 20a). The actual cross sectional area at the peak compressive load was 0.0505 square inches, which gave the true stress at the peak compressive load as approximately 245 ksi. The actual cross sectional area at the peak tensile load was 0.0466 square inches. This gave the true stress value at the peak tensile load as 249 ksi which was almost identical as the true stress at peak compressive load.

2. Plastic Strain Behavior

From the results shown in Figs. 21-24 it is obvious that Coffin-Manson's law is a good approximation to the low cycle fatigue behavior of TRIP steels. These steels exhibited better low cycle fatigue

properties at 200°C than that at room temperature. Since 200°C was above the M_D temperature of these alloys, there was no martensitic transformation, while at room temperature cycling there was transformation taking place. Hence these results indicated that the transformation did not give improved low cycle fatigue properties although it gives better uniform elongation. It has been reported elsewhere¹⁻⁵ that the transformation inhibits necking during tensile testing thereby giving higher values of tensile elongation at room temperature than at 200°C. In the Coffin-Manson law, the tensile fracture ductility is taken based on initial and final area of cross section (i.e., % reduction in area). Hence the increased uniform elongation need not give better low cycle fatigue properties. The percentage reduction in area at 200°C was higher than that at 25°C and hence better low cycle fatigue properties at 200°C were obtained.

In fact it seems that the more stable the steel is, the better the low cycle fatigue properties are. The comparison between LCF curves for the alloy 80% deformed at 450°C and the alloy 80% deformed at 250°C also illustrates this (Fig. 67). The alloy deformed 80% at 450°C was less stable than the one deformed by the same amount at 250°C under cyclic loading as shown by the difference in the amount of martensite in the final fracture area (Fig. 37). The alloy 80% deformed at 250°C as expected showed better low cycle fatigue properties than the one deformed 80% at 450°C. The 80% deformed alloys were much more unstable than 20% deformed ones as discussed in the section on tensile results and also shown by martensitic measurements in the fracture area. Hence 20% deformed alloys should give better low cycle fatigue properties than the 80% deformed ones,

if the above interpretation is valid. Figures 68 and 69 clearly demonstrate the better low cycle fatigue properties of 20% deformed alloys thereby supporting the above conclusion. Another factor to be examined for the validity of expression (2), is how well tensile fracture ductility fitted the LCF curve. This can be best investigated by comparing tensile fracture ductility ϵ_f with fatigue ductility. The fatigue ductility of a material is obtained by the extrapolation of LCF curve to $N = 0.25$. Figure 70 shows the relationship between tensile and fatigue ductility. Except for 200°C tests, the agreement was fairly good considering the statistical nature of fatigue.

The use of this fatigue ductility in place of tensile fracture ductility ϵ_f , in the relationship

$$N_f^{1/2} \epsilon_{PR} = 1/2 (\epsilon_f - \epsilon_0) \quad (37)$$

has been suggested by Sach et al.⁵⁰ In the Figs. 25-28, $N = 0.25$ points based on fatigue ductility are shown by an 'X' mark. As can be seen from these marks use of fatigue ductility gave a better agreement with expression than tensile fracture ductility.

3. Fracture Behavior

The room temperature low cycle fatigue failures for the alloy deformed 20% at 450°C were essentially ductile as evidenced by the predominance of dimples in the fractographs from the fracture surfaces of different specimens (Figs. 58 through 59). If the features in Figs 45(e) are taken as striation marks, one gets an approximate crack propagation rate of 5×10^{-5} in/cycle. This is obtained based on the assumption that one striation spacing corresponds to crack growth in one

cycle.^{51,52} This specimen had failed after 98 cycles only, hence these marks could not be striations. These marks were probably due to rubbing and are formed after the crack has passed from repeated opening and closing of an asperision on the opposite surface during fatigue crack propagation.

It can be seen in Fig. 44, (158 cycles to failure) that the dimple size in low cycle fatigue is approximately 1/10 of the average dimple size in tensile fracture which is similar to the findings for fatigue crack propagation under high stress intensity conditions. The tire tracks observed in Fig. 46 are usually associated with high stress, low cycle fatigue. These marks are thought to result from the relative motion between two closely mating fracture surfaces under the action of high stress.

In Fig. 45(b) both dimpled rupture and flat regions are noted. The flat region sandwiched between the dimpled region is thought to be due to the fracture in martensite induced in the parent austenite matrix due to plastic strain cycling. This type of change in fracture mode in TRIP steels has also been observed.

The 200°C fracture surfaces do show evidence of a large amount of plastic deformation, serpentine glide and considerable degree of stretching (Fig. 44). The fractograph in Fig. 44(a) exhibits a surface which probably consisted of a series of striations that may have resulted from an extensive slip of grain boundary sliding during rupture. Also there is some evidence of serpentine glide in Fig. 44(a) and stretching in Fig. 44 (b). Serpentine glide is postulated to be formed by slip (glide) along planes intersecting a free surface during the straining process, where stress is acting essentially parallel to the surface.

These smooth surface features, because of the associated plastic deformation and lack of ductile features, have also been called "ductile cleavage" by some investigators^{53, 54}.

The uniformly elongated dimples as shown in Fig. 41 are characteristic of a fast crack propagation rate. Gerberich⁵⁵ et al. have also reported them in high strength steels and associated them with a high stress-intensity range and propagation rate. Since the specimen had failed only after 158 cycles the high crack propagation rate was expected as evidenced by the elongated dimples.

c. Fatigue Crack Propagation

1. Crack Growth Behavior

The loading used in the present investigation was periodic. The specimen was always brought to zero load after each run for crack-length measurements prior to starting the periodic load for the next run. No effect due to this interruption was noted. Other investigators have also observed no effect of any consequence.^{15,16} The crack growth continues as if no interruption had occurred. Moreover, if the load range (or ΔK) is increased, the crack growth rate observed is characteristic of the higher load. However, when a shift from a large to a small load range is made, the crack growth stops as is illustrated by run numbers 13-15, for specimen S-22 in Table XVIII. The delays after the overload are explained as being caused by either the residual compressive stresses near the crack tip⁵⁶ or a blunting of the crack-tip by large deformations associated with the high load or a change in plastic zone size,¹⁶ or perhaps all three. Hence, in the present investigation almost all the runs on any one specimen were made in such a way that the load range for one particular run was either more, or the same, as the previous run.

It has been mentioned earlier that the frequency of cycling was usually in the range of 1 cps to 4 cps. This range of cyclic speed is not expected to have an effect of any consequence. On these steels Wei, et. al.⁵⁷ have reported that for maraging steels, cyclic speeds in the range of 4 cps to 150 cps have little influence on the rate of fatigue crack propagation.

Expression (34) developed for TRIP steels indicates that besides the fourth power of stress-intensity range, ΔK , the higher powers of ΔK

also should have an effect on the rate of crack propagation. However, as can be seen from Figs. 52-55, no significant effect of the higher powers of ΔK on crack propagation rate was observed. It will be shown now why no such contribution of higher powers of ΔK was observed in the present investigation, where $\Delta\sigma/\sigma_{ys}$ was usually below 0.2 and a/w was always between 0.34 and 0.65.

Expression (34) can be written as

$$\frac{da}{dn} = c^2 \left(\frac{\Delta K}{\sigma_{ys}} \right)^4 \left[1 + \frac{2D}{c} \left(\frac{\Delta K}{\sigma_{ys}} \right)^2 + \frac{D^2}{c^2} \left(\frac{\Delta K}{\sigma_{ys}} \right)^4 \right] \quad 34a$$

For $a/w = 0.34$ and $\Delta\sigma/\sigma_{ys} = 0.2$ the above expression after solution and simplification reduces to

$$\frac{da}{dn} = c^2 \frac{\Delta K}{\sigma_{ys}} [1 + 2 \times 10^{-2} + 8 \times 10^{-5}] \quad (38)$$

while for $a/w = 0.65$ and $\Delta\sigma/\sigma_{ys} = 0.2$, the expression (34a) reduces to

$$\frac{da}{dn} = c^2 \frac{\Delta K}{\sigma_{ys}} [1 + 3 \times 10^{-2} + 2 \times 10^{-4}] \quad (39)$$

In both expressions 38 and 39, the second and the third terms in the brackets are the contributions of the sixth and eighth power terms $(\Delta K/\sigma_{ys})$ to the crack propagation rate. It can be seen that this contribution is not significant. However, at higher values of $\Delta\sigma/\sigma_{ys}$ some contribution due to higher order terms is expected. Indications of this can be noted in Figs. 52 and 55, (for the alloy 20% deformed at 450°C and the alloy deformed 80% at 450°C), where at lower ΔK values, the data are more or less uniformly above or below the FCP* curve while at higher ΔK values, they are usually below the FCP curve. At this stage because of the lack of sufficient data points at higher range of ΔK , no conclusive statement can be made.

* FCP curve henceforth refers to the log-log plot of stress-intensity range, ΔK , as a function of crack propagation rate da/dn .

Since, in the present investigation, the higher order terms in expression (34) did not have any effect, the expression (34) can be simplified and an expression similar to (27) can be obtained

$$\frac{da}{dn} = \frac{8}{l} \left(\frac{\epsilon_y}{\epsilon_f} \right)^2 \left(\frac{\Delta K}{\sigma_{ys}} \right)^4 \quad (40)$$

ϵ_y , ϵ_f and σ_{ys} were obtained from the tensile data in Table V.

The rate of crack growth, da/dn was determined at $\Delta K = 50 \text{ ksi} \cdot \sqrt{\text{in.}}$ for each case from the corresponding FCP curve. The values of l calculated from these data are listed below

Processing	da/dn at 50 ksi $\text{in}^{1/2} \times 10^5 \text{ in/cycle}$	$l \times 10^4 \text{ in.}$
20% deformation at 450°C	6.40	26.2
20% deformation at 250°C	15.80	9.5
80% deformation at 250°C	6.25	9.7
80% deformation at 450°C	7.50	10.3

As was mentioned earlier, ' l ' was the region in which the fracture criterion (expression (21)) had to be satisfied. The values of ' l ' for the four cases investigated are within a factor of three and are of the same order of magnitude as a typical cell or subgrain size. Hence the magnitude ' l ' is apparently related to the substructure of the material and can be taken as a material property. The values of l for the four treatments show that there is no obvious correlations between the "structural size" l and the rate of crack propagation, da/dn .

Expression (40) can be written as

$$\frac{da}{dn} = R (\Delta K)^4 \quad (41)$$

where R is a rate constant. The magnitude of R represents the ease of fatigue crack propagation in a material. The relative magnitude of R indicates the sensitivity of the rate of fatigue crack propagation to

the different processing treatments. The values of R (based on FCP curve) are tabulated below

Processing	$R \times 10^{23}$ $\text{in}^7 \text{lb}^{-4} \text{cycle}^{-1}$	Critical Stress-Intensity Factor K_{IC} , ksi-in. ^{1/2}
20% deformation at 450°C	1.02	331
20% deformation at 250°C	2.55	279
80% deformation at 250°C	1.00	296
80% deformation at 450°C	1.2	334

Although no definite conclusion could be drawn from the above results, it appeared that the treatment with the lowest value of K_{IC} (20% deformation at 250°C) had the highest value of R, i.e., maximum ease of fatigue crack propagation. Since in the present case the plain strain fracture toughness parameter K_{IC} was not determined, nothing could be inferred as to whether there is any correlation between the K_{IC} and da/dn as suggested by Kraft.⁵⁸ However, based on K_{IC} data the present results did not indicate any obvious relationship. Wei⁵⁷ et al have also found no relationship for medium-carbon low alloy ultrahigh strength steel and maraging steels in an inert environment. The R value can be used to grade different materials as well as various thermal, mechanical and thermomechanical treatments with respect to their fatigue crack propagation properties. Also the value of R can be used as a design criterion, at least, for TRIP steels as the data agreed reasonably well with the fourth power relationship. Another fact which increases the usefulness of the R values obtained, as design criteria is the lack of a pronounced thickness effect in the crack growth rate behavior, as reported by Donaldson and Anderson⁵⁹ and others^{52,57} for several alloys. Most of these alloys show a much more significant thickness effect on toughness values as compared to TRIP steels.⁶⁰ Hence

a useful conclusion is that the crack growth rate behavior of TRIP steels (for the composition and treatments investigated) that was observed for a moderate thickness (0.075") approximately represents the behavior for other thicknesses.

2. Comparison with Other Materials

To compare the fatigue crack propagation behavior of 'TRIP' steels with that of the other materials, two groups of alloys were selected; one group with an ultimate tensile strength (UTS) in the vicinity of 250 ksi (for comparison with 80% deformed alloys, Fig. 71) and the other groups with a UTS of approximately 185 ksi (for comparison with 20% deformed alloys, Fig. 72). The first group consisted of four materials: 18 Ni (250) maraging steels, 300 M steel, AM 355 CRT stainless steel and PH 15-7 Mo steel. The second group consisted of AM 350 CRT stainless steel. The sources of these data are given in the references indicated in the figure. The ultimate tensile strengths of the materials are also shown in the figures.

It can be seen that the steel deformed 80% at 450°C is better than the 300 M AM355 CRT and PH 15-7 Mo steels and also compares favorably with maraging steels in the low stress intensity range. Wei⁵⁷ et al. have reported a value of 1.6 for R in the case of maraging steels in the low stress-intensity range, while for the alloy deformed 80% at 250°C, R was 1.00. This indicates the higher resistance of this TRIP steel to fatigue crack propagation than maraging steels.

At higher values of ΔK , as is discussed in the next section the formation of a comparatively large amount of martensite (higher in C-content compared to the martensite in maraging steels) decreases the fatigue crack propagation resistance of TRIP steels, with the result that maraging steels appear to be superior

3. Fracture Behavior

From Figs. 59 through 62, it seems that fracture appearance of the fatigue crack specimens of TRIP steels comprised of fatigue striations, quasi-cleavage and elongated dimples. Although fatigue striations were observed in both 20% and 80% deformed alloys, they were not always well-developed and present. The absence of well developed striations as well as lack of their abundance is quite common in high strength steels.⁶² However, the striations were observed on a portion of fatigue crack surfaces perpendicular to the tensile direction. Several investigators have discussed the mechanisms of striation formation based on blunting and resharpening of the crack-tip during load cycling⁶³ and according to these mechanisms each striation is associated with one load cycle. Hence, as mentioned earlier, the striation spacing can be taken as a measure of the crack propagation rate. Much effort has been devoted to the study of the relationship between the striations and the fatigue crack growth rate. For high strength aluminum alloys a reasonable agreement has been reported between striation spacing and crack growth rate.⁵²

Some of the striation spacings observed and the corresponding crack propagation rates for TRIP steels are listed below.

Processing	Striation Spacing $S \times 10^5$ in.	Crack Propagation Rate* $da/dn \times 10^5$ in.
20% at 450°C	1.48	1.00
20% at 450°C (Fig. 59a)	1.68	1.00
20% at 450°C (Fig. 59b)	19.7	24.03
80% at 450°C (Fig. 61a)	0.785	0.62

Processing	Striation Spacing $S \times 10^5$ in.	Crack Propagation Rate [*] $da/dn \times 10^5$ in.
80% at 450°C	0.79	0.62
80% at 450°C (Fig. 61b)	1.48	1.78
80% at 450°C	1.73	1.78
80% at 450°C	1.83	1.78
80% at 450°C	1.97	3.94
80% at 450°C	2.10	3.94

From this it can be seen that as da/dn increases, the spacing also increases. Since, the macroscopic crack propagation rate, da/dn , corresponds to the striation spacing, the striation spacing should also vary in the same manner as the macroscopic crack propagation rate does with the stress intensity range. The results of the present investigation were broadly consistent with such a dependence as can be seen from Figs. 52 and 55, where microscopic crack propagation rates (striation spacing) are indicated by filled circles. The band has been drawn for uncertainty in the ΔK values. This correlation is of significant importance for post-fracture analysis of engineering service failures. Towards the later stages of crack propagation, i.e. at high values of stress intensity range, elongated dimples were observed (Fig. 61c). The elongated dimples are associated with the change in the fracture mode from flat to shear type.⁵² The dimples were oriented with the long axis perpendicular to the crack growth direction. The rate of crack propagation corresponding to the fractograph in Fig. 61c was 6.3×10^{-5} in/cycle. The occurrence of these dimples at high crack propagation rates has already been discussed.

* The da/dn values listed are the average of the da/dn values in the fracture surface from which the replica was obtained.

Somewhat similar and more uniformly elongated dimples were also observed in low-cycle fatigue fracture (Fig. 41). Figure 60 (obtained from the fracture surfaces of the alloy deformed 20% at 450°C) and Fig. 62b (obtained from the fracture surface of the alloy deformed 20% at 450°C) exhibit features of brittle failure. This was not unexpected, since besides austenite, the fracture surface contained martensite as was shown by low cycle fatigue results (Fig. 37). In Fig. 60 quasi-cleavage areas can be observed along with a region resembling fatigue striations. Since the spacing of these "fatigue striations" is approximately one-hundredth of the crack propagation rate (spacing = 5.9×10^{-6} in 1 cycle and $da/dn = 6.57 \times 10^{-4}$ in/cycle), these 'striations' are just rubbing marks similar to the 'tire tracks' observed in Fig. 42 for the low cycle fatigue fracture surface. Features resembling river markings can also be seen in Fig. 60 and these are always associated with cleavage fracture. The featureless fractograph in Fig. 62b (from the fatigue fracture surface of the alloy deformed 80% at 450°C) is believed to be due to fracturing of martensite induced in the parent austenite during load cycling. In Fig. 62a a structure resembling "brittle striations" can be seen. This is usually obtained in high strength alloys²⁶ and in the present case these could be due to the presence of martensite on the fracture surface. The existence of brittle striations can also be inferred from the appearance of the crack in Fig. 58(a). Here small cracks branching from the main crack are shown. Forsyth⁶⁴ and Laird⁶³ have discussed the formation of such branching cracks in detail. According to them, these cracks are formed by the cleavage fracture which subsequently and in the course of the tension part of the load cycle,

changes to ductile shear fracture. Based on their model, the spacing between successive branched cracks can be associated with each stress cycle. Hence the spacing between the two branches should be approximately equal to the crack propagation rate if only cleavage striations are forming. In the present case, the spacing was roughly of the order of $10(da/dn)$ which seems to disallow the possibility of cleavage striations. However, in TRIP steels both magnetic measurements (for low cycle fatigue fracture specimens) and fractographic results indicated that the fracture area comprised of austenite as well as martensite the former being much greater in amount. As a result the major part of the crack propagation is by ductile fracture while the remainder is by cleavage fracture. Hence the branch-spacing observed is justified. In fact this branching behavior is a further confirmation of the previous observation that the fracture surface consists of both ductile and cleavage features.

D. Role of Martensitic Transformation on the Behavior
of TRIP Steels Under Cyclic Loading

The austenite to martensite transformation in TRIP steels has been shown to play a beneficial role on both the tensile and fracture toughness properties.^{1,3} This beneficial effect of the martensitic transformation on uniform elongation is due to the inhibition of necking by strain-hardening, while the toughness is enhanced due to the absorption of energy by the martensitic transformation occurring at the crack-tip. However, this transformation does not seem to play a similar beneficial role on cyclic properties. Low cycle fatigue properties were better in the absence of martensitic transformation. In the case of fatigue crack propagation properties no beneficial effect of the transformation was observed.

Let us first consider the role of martensitic transformation on the low cycle fatigue behavior of TRIP steels. For this purpose the number of cycles to fracture is plotted against percent martensite in the fracture area for various cyclic plastic strain ranges in Fig. 73. From this figure, it is clear that for a given value of strain range, the smaller the amount of martensite formed, the better was the life obtained. Apparently the martensite formed gave rise to more crack initiation sites and an easy crack propagation path in the steel.

Fatigue crack propagation can be considered as the result of cumulative damage caused by strain cycling to the material at the crack tip. For crack propagation in a load-cycled specimen, the material ahead of the crack tip is strain cycled and experiences increasing amplitudes of strain cycling as the crack propagates toward that point. The cumulative strain could be very high. The high cyclic plastic strain range

induces a large amount of martensite in an austenite matrix and this increases the strain hardening rate. It has been shown by Cotterell⁶⁵ that a high strain hardening rate gives rise to a high crack propagation rate. Furthermore, the formation of martensite in ductile austenite decreases the amount of strain accumulation required to cause fracture (ϵ_f in Expression 21). Hence, no beneficial effect of martensitic transformation on cyclic properties was observed.

VI. SUMMARY AND CONCLUSION

The purpose of the present investigation was to study the behavior of TRIP steels under cyclic loading. In the present study, an alloy of such a composition was chosen as to have the M_s temperature well below room temperature, while the M_D temperature was above room temperature after thermo-mechanical processing. High-strain low cycle fatigue experiments on round and well polished specimens as well as fatigue crack propagation tests on single edge notched specimens were carried out. The diametral strain was maintained constant during low-cycle fatigue tests. Under this condition, life of the specimen is related to the plastic strain range by the well known Coffin -Manson law. The effect of martensitic transformation (which takes place in this steel) on this law was investigated for this alloy after various thermo-mechanical treatments. For this purpose room temperature low-cycle fatigue tests were run at different cyclic strain levels. To study the effect of mixed austenite-martensite matrix, tests were also done after very high prestrain. This high value of prestrain induced martensite in the austenite matrix of TRIP steel. The amount of martensite induced was magnetically measured by a "permeameter" built specifically for this purpose. In order to establish the role played by the martensitic transformation on low cycle fatigue, tests were also run at 200°C. Since 200°C was above the M_D for this alloy, no transformations took place during cycling. The fracture surfaces of low-cycle fatigue specimens were observed using scanning and transmission electron microscopes.

Fatigue crack propagation tests at various ΔK levels were run on single edge notched specimens. A simple theoretical model of fatigue

crack propagation based on experimental observations was developed. The picture of fatigue crack growth presented by the model is as follows. In notched plates, the material ahead of a crack-tip is strain cycled. On cycling, the strain keeps accumulating in a small region ahead of the crack-tip. Soon the fracture criterion for this region is satisfied and the crack propagates. The fracture criterion is satisfied throughout the propagation.

The replicas of the fracture surfaces of the SEN specimens were observed in a transmission electron microscope.

The experimental results may be summarized as follows:

(1) The low cycle fatigue life of TRIP steels both at room temperature (in the presence of martensitic transformation) and at 200°C (in the absence of the transformation) and at 200°C (in the absence of the transformation) was found to be related to the plastic strain range by the Coffin-Manson relationship.

$$\epsilon_{PR} N_f^{1/2} = C$$

(2) The monotonic fracture strain (obtained from reduction in area measurements), ϵ_f can be used as plastic strain range, ϵ_{PR} at $N = 0.25$ for room temperature tests as the room temperature LCF curve extrapolated reasonably well to the tensile ductility (the alloy deformed 80% at 450°C was an exception).

(3) The strain cycling behavior of prestrained specimens showed the following relationship

$$\epsilon_{PR} N_f^{1/2} = 1/2 (\epsilon_f - \epsilon_0), \text{ where,}$$

ϵ_0 is the prestrain

(4) For all the four processing treatments investigated, the alloy showed better low cycle fatigue properties at 200°C, thereby indicating that the absence of martensitic transformation was beneficial. However when specimens with and without martensitic transformation are compared at

the same stress amplitude, the steel undergoing martensitic transformation is always superior in that it can withstand a large number of stress cycles before failure.

(5) Either cyclic hardening or softening occurred at room temperature depending primarily upon the strain range used in cycling. Whenever plastic strain range was greater than approximately 3% in amplitude, there was hardening, while for plastic strain range below 3% there was softening.

(6) For 200°C low cycle fatigue tests, only cyclic softening was observed in all the cases.

(7) The amount of martensite measured in the fracture area increased with the cyclic plastic strain range in each case.

(8) Depending on the number of cycles to failure, the room temperature low-cycle fatigue fracture surfaces showed rupture type dimples, rubbing marks ("tire tracks") uniformly elongated dimples and striations.

(9) The fractographs obtained from 200°C fatigue fracture surfaces showed evidence of a large amount of plastic deformation, serpentine glide and considerable degree of stretching.

(10) To a first approximation, the experimental results were found to be in agreement with the theoretical model and showed the correlation between the stress-intensity range and fatigue crack propagation rates as

$$da/dn = R (\Delta K)^4$$

(11) The rates of fatigue crack propagation for three of the four processing treatments investigated were about the same and were relatively

insensitive to strength, ductility and toughness.

(12) A sudden shift from a large to a small range of load variation was found to delay the crack growth.

(13) In fatigue crack propagation tests small cracks branching from the main cracks were observed. These branching cracks could be associated with cleavage fracture.

(14) No obvious correlation between the "structural size", l (the region in which the fracture criterion for crack growth has to be satisfied) and the rate of crack-propagation, da/dn was observed.

(15) The fracture appearance of the fatigue-crack specimens of TRIP steels, comprised of fatigue striations, quasi-cleavage and elongated dimples reflecting the extremely complex structure of TRIP steels.

(16) There seems to exist a correlation between the striation spacing and the stress intensity range. This correlation may be of importance for post fracture analysis of engineering service failures.

(17) The alloy deformed 80% at 250°C showed the best fatigue-crack propagation properties of all the four processing treatments investigated. It is better than a number of high alloy steels of similar tensile strength levels and compares favorably with maraging steels in the low range of ΔK .

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APPENDIX A

According to Irwin, the plastic zone size for unidirectional loading is given as:

$$R_p^* = \frac{k^2}{\pi \sigma_{ys}^2}$$

On the basis of this, the plastic zone size for repeated loading is given as

$$R_p = \frac{(\Delta k)^2}{4\pi \sigma_{ys}^2}$$

This statement can be understood by referring to Fig. A-1. Here a cracked body is loaded by a system of stresses proportional to some parameter be reduced by an amount ΔP . In order to consider the changes in stress and strain distributions, it can be looked as loading by an amount ΔP in opposite direction. Here also the stress-intensity factor is effectively infinite at the crack-tip, and hence reverse plastic flow commences with the first increment of load reduction.

This creates a new plastic zone of reversed deformation imbedded in the plastic zone accompanying the original loading. Now, this change in strain and displacement due to the load reduction are determined by ΔP and $2\sigma_{ys}$ in the same way as the strain and the displacement during monotonic loading are determined by P and σ_{ys} .

Similarly, for reloading $P - \Delta P$ to P and subsequent load cycles, the load fluctuation ΔP will determine the size of the plastic zone. Thus, the plastic zone size under repeated loading has been obtained simply by making the following changes in the plastic zone size under monotonic loading:

- (1) Replacing the loading parameter P by ΔP
- (2) Doubling the yield stress.

TABLE I. Chemical composition of the alloys used for low-cycle fatigue tests

Alloy Identification No.	Ingot No.	Cr	Ni	Mn	Si	Mo	C	Fe
A	6711-4	9.5	7.6	3.0	2.8	4.0	0.26	balance
B	6711-6	9.5	7.7	3.0	2.8	4.0	0.27	balance
C	682-1	9.0	7.5	2.9	2.8	4.0	0.26	balance
D	682-2	9.0	7.5	2.9	2.8	4.0	0.25	balance

TABLE II. Chemical composition of the alloys used for fatigue crack propagation tests.

Alloy Identification No.	Ingot No.	Cr	Ni	Mn	Si	Mo	C	Fe
a ₁	697-1A	9	7.4	2.8	2.8	4.0	.24	balance
a ₂	701-1A	9	7.4	2.8	2.8	4.0	.25	balance
b ₁	697-19A	9	7.5	2.9	2.8	4.0	.25	balance
b ₂	698-1A	9	7.5	2.9	2.8	4.0	.25	balance
c ₁	697-19B	9	7.5	2.9	2.8	4.0	.24	balance
c ₂	698-1B	9	7.5	2.9	2.8	4.0	.25	balance
d ₁	695-3A	9	7.4	2.9	2.8	4.0	.24	balance
d ₂	695-3B	9	7.4	2.9	2.8	4.0	.24	balance
d ₃	697-1B	9	7.4	2.8	2.8	4.0	.24	balance

TABLE III. Tensile properties of round (low cycle fatigue)specimens at room temperature after various thermochemical processing

Alloy Identi- fication No.	Processing	Y.S. 1000 psi	T.S. 1000 psi	Elonga- tion %	ϵ_f , true fracture strain %	Reduction in area %	Magnetic Characteristics*	
							Before Test	After Test
A	20% at 450°C	122	158	52	57	44	A	M
B	20% at 250°C	135	184	63	68	52	A	M
C	50% at 250°C	243	243	42	60	46	A	M
D	30% at 450°C	228	237	36	26	24	A	M
* M = magnetic, A = non-magnetic								

TABLE IV. Tensile properties of round (low cycle fatigue) specimens at 200°C after various thermo-mechanical processing.

Alloy Identi- fication No.	Processing	Y.S. 1000 psi	T.S. 1000 psi	Elonga- tion %	ε _f true fracture strain %	Reduction in area %	Magnetic Characteristic *	
							Before Test	After Test
A	20% at 450°C	120	138	28	86	56	A	A
B	20% at 250°C	115	131	36	87	58	A	A
C	80% at 250°C	225	225	20	87	58	A	A
D	80% at 450°C	215	215	20	80	55	A	A

*
A = non-magnetic

TABLE V. Summary of Stress-Intensity and Tensile Data at 25°C for 0.075 Inch Thick TRIP Steel Plate after Various Thermo-Mechanical Processing

Alloy Identification No.	Processing	Y.S. 1000 psi	T.S. 1000 psi	Elongation %	Stress-Intensity (ksi-in ^{1/2})	Magnetic*	
						Before Test	After Test
a ₁	20% at 450°C	137	183	46		A	M
a ₂	20% at 450°C	138	187	42	331**	A	M
b ₁	20% at 250°C	126	174	55		A	M
b ₂	20% at 250°C	119	160	49	279	A	M
c ₁	80% at 250°C	237	258	36		A	M
c ₂	80% at 250°C	237	237	38	296	A	M
d ₁	80% at 450°C	233	248	34		A	M
d ₂	80% at 450°C	231	243	32	334**	A	M
d ₃	80% at 450°C	238	260	36		A	M

* A = Non-magnetic M = Magnetic

** Tested at a cross-head speed of 0.2"/sec. At 0.02"/sec. crack grew slowly giving no instability.

TABLE VI. Summary of room temperature cyclic data of Alloy A after 20% deformation at 450°C.

Specimen No.	Total Diametral Strain Range % ϵ_d^+	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
305	7.55	315	6.92	7.97	32	A	M
304	5.32	303	4.72	5.72	64	A	M
303	3.65	277	3.10	4.02	124	A	M
302	2.23	242	1.75	2.55	508	A	M
306	1.28	195	0.89	1.54	1356	A	M

* Refer to figure 10

** M = magnetic M¹ = Feebly magnetic A = Non-magnetic

*** Estimated in the fracture area only

TABLE VII. Summary of room temperature cyclic data of Alloy B after 20% deformation at 250°C.

Specimen No.	Total Diametral Strain Range % ϵ_c	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
206	7.70	363	6.97	8.18	25	A	M
205	5.56	309	4.94	5.97	46	A	M
203	3.78	266	3.25	4.13	81	A	M ¹
202	2.47	253	1.96	2.81	276	A	M
204	1.52	214	1.10	1.81	771	A	M ¹
207	1.28	211	0.85	1.56	1460	A	M ¹

* Refer to fig. 10

** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only.

TABLE VIII. Summary of room temperature cyclic data of alloy C after 80% deformation at 250°C.

Specimen No.	Total Diametral Strain Range, % ϵ_d^+	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
107	7.71	484	5.74	8.36	22	A	M
106	7.60	481	5.54	8.24	26	A	M
105	5.96	458	5.04	6.57	32	A	M ¹
104	5.84	442	2.96	4.43	70	A	M ¹
102	2.54	411	1.72	3.09	220	A	M ¹
101	1.51	344	0.82	1.97	419	A	M ¹
103	1.28	329	0.62	1.72	649	A	M ¹
108	1.13	298	0.53	0.93	1579	A	M ¹

* Refer to Fig. 10

** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only.

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TABLE IX. Summary of room temperature cyclic data of Alloy D after 80% deformation at 450°C

Specimen No.	Total Diametral Strain Range, % ϵ_d	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
16	8.00	502	7.00	8.66	12	A	M
17	5.90	492	4.92	6.56	22	A	M
18	4.91	468	3.97	5.53	32	A	M
19	3.87	442	2.99	4.36	98	A	M
21	2.88	424	2.03	3.45	146	A	M
24	2.50	403	1.69	3.04	171	A	M ¹
25	2.50	415	1.67	3.05	176	A	M ¹
26	2.50	416	1.67	3.05	179	A	M ¹
27	1.50	368	0.76	1.99	312	A	M ¹
29	1.50	360	0.78	1.98	321	A	M ¹
32	1.50	334	0.83	1.95	376	A	M ¹
30	1.50	267	0.97	1.81	381	A	M ¹
33	1.25	326	0.60	1.68	873	A	M ¹
31	0.88	287	0.35	1.25	1514	A	M ¹

*Refer to Fig. 10

** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only.

TABLE X. Summary of room temperature cyclic data of Alloy A after 20% deformation at 450°C and a prestrain of approximately 21.5%.

Specimen No.	Amount of Prestrain %	Total Diametral Strain Range ϵ_d	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
			Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test (After Pre-strain)	After Test***
357	21.50	4.85	390	4.07	5.37	40	M ¹	M
355	21.50	4.85	370	4.11	5.34	44	M ¹	M
358	21.75	3.95	344	3.26	4.41	50	M ¹	M
354	21.50	3.97	343	3.27	4.43	75	M ¹	M
356	21.75	3.00	333	2.33	3.44	94	M ¹	M
353	21.75	2.96	343	2.26	3.42	104	M ¹	M
352	21.50	1.94	306	1.33	2.35	181	M ¹	M
351	21.50	1.24	256	0.73	1.58	812	M ¹	M

* Refer to Figure 10

** M = Magnetic M¹ = Feebly Magnetic A = Non-magnetic

*** Estimated in fracture area only.

TABLE XI. Summary of room temperature cyclic data of Alloy B after 20% deformation at 250°C and a prestrain of approximately 21.5%.

Specimen No.	Amount of Prestrain %	Total Diametral Strain Range ϵ_d^+	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
			Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test (After Pre-strain)	After Test***
255	21.50	4.95	485	3.98	5.60	25	M ¹	M
259	21.50	4.85	432	3.98	5.43	27	M ¹	M
254	21.50	4.00	372	3.26	4.50	72	M ¹	M
258	21.50	4.00	408	3.18	3.73	72	M ¹	M
257	21.50	2.94	378	2.18	3.44	98	M ¹	M
256	21.75	2.96	332	2.30	3.40	213	M ¹	M
252	21.5	1.95	322	1.31	2.38	343	M ¹	M
261	21.5	1.27	268	0.73	1.60	592	M ¹	M
251	21.5	1.26	284	0.69	1.64	709	M ¹	M

TABLE XII. Summary of room temperature cyclic data of Alloy C after 80% deformation at 250°C and a prestrain of approximately 11%.

Specimen No.	Amount of Prestrain %	Total Diametral Strain Range ϵ_d^+	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
			Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test (After Pre-strain)	After Test***
155	10.75	4.70	532	3.64	5.41	24	M	M
154	10.5	3.75	525	2.70	4.45	59	M	M
153	10	2.82	512	1.75	3.50	97	M	M
152	10.5	1.87	396	1.08	2.40	367	M	M
151	10.75	1.18	282	0.62	1.56	500	M	M

TABLE XIII. Summary of room temperature cyclic data of Alloy D after 80% deformation at 450°C and a prestrain of approximately 11%.

Specimen No.	Amount of Prestrain %	Total Diametral Strain Range ϵ_d^+	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
			Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
							(After Pre-strain)	
56	10.75	4.70	558	3.58	5.44	7	M	M
57	10.75	4.66	600	3.46	5.46	20	M	M
54	10.75	3.74	566	2.61	4.49	32	M	M
55	10.75	3.14	538	2.06	3.86	52	M	M
52	10.75	2.38	492	1.40	3.04	75	M	M
53	10.50	1.90	425	1.05	2.47	219	M	M
51	11.25	1.19	310	0.57	1.60	396	M	M

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TABLE XIV. Summary of 200°C test temperature cyclic data of Alloy A after 20% deformation at 450°C

Specimen No.	Total Diametral Strain Range $\% \epsilon_d^+$	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
746	10.5	245	10.01	10.83	71	A	A
744	6.35	226	5.80	6.72	145	A	A
742	4.65	240	4.17	4.97	157	A	A
745	4.30	211	3.88	4.58	365	A	A

* Refer to Fig. 10

** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only.

TABLE XV. Summary of 200°C test temperature cyclic data of Alloy B after 20% deformation at 250°C

Specimen No.	Total Diametral Strain Range % ϵ_d^+	Stable Values *			Cycles to failure N_f	Magnetic Characteristics **	
		Stress Range ksi $\Delta\sigma$	Elastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test ***
647	13.6	254	13.09	13.94	39	A	A
646	8.25	241	7.77	8.57	94	A	A
645	5.2	218	4.76	5.49	188	A	A
648	3.7	228	3.24	4.00	401	A	A
644	2.08	194	1.70	2.34	3163	A	A

* Refer to Fig. 10

** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only

TABLE XVI. Summary of 200°C test temperature cyclic data of Alloy C after 80% deformation at 250°C.

Specimen No.	Total Diametral Strain Range % ϵ_c^+	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
544	12.40	405	11.59	12.94	2.9	A	M ¹
543	9.50	400	8.70	10.03	41	A	M ¹
541	3.10	367	2.47	3.59	179	A	M ¹
542	2.50	317	1.87	2.92	1060	A	M ¹

* Refer to Fig. 10

** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only

TABLE XVII. Summary of 200°C test temperature cyclic data of Alloy D after 80% deformation at 450°C.

Specimen No.	Total Diametral Strain Range % ϵ_d	Stable Values*			Cycles to failure N_f	Magnetic Characteristics**	
		Stress Range ksi $\Delta\sigma$	Plastic Strain Range, % ϵ_{PR}	Total Strain Range, % ϵ_{TR}		Before Test	After Test***
444	6.10	385	5.33	6.61	3.2	A	A
445	4.40	366	3.67	4.89	4.5	A	A
443	2.80	350	2.10	3.27	215	A	A
441	1.510	320	0.87	1.94	460	A	A
442	1.260	292	0.68	1.94	1346	A	A

* Refer to Fig. 10

*** M = Magnetic M¹ = Slightly magnetic A = Non-magnetic

*** Estimated in fracture area only

Table XVIII. Fatigue crack propagation test data of alloys after 20% deformation at 450°C.

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate $\frac{da}{dN}$ (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P_{max}	Minimum P_{min}	Start	End	Average
S-22 Alloy a ₁	2	4	10,000	0.6890	0.6980	0.090	940	100	14.83	15.23	15.03
	3	4	15,000	0.6980	0.7114	0.089	940	100	15.23	15.63	15.43
	4	4	15,000	0.7114	0.7264	0.100	930	90	15.63	16.01	15.82
	5	4	15,000	0.7264	0.7489	0.150	930	90	16.02	16.77	16.39
	6	4	20,000	0.7489	0.8043	0.277	930	90	16.77	18.66	17.73
	7	4	10,000	0.8043	0.8186	0.143	930 - 855*	90	18.65	18.35	18.50
	8	4	10,000	0.8186	0.8306	0.120	920	85	19.03	19.66	19.34
	9	4	10,000	0.8306	0.8549	0.243	920	95	19.42	20.31	19.87
	10	4	10,000	0.8549	0.8685	0.136	920	95	20.31	20.73	20.52
	11	4	10,000	0.8685	0.8769	0.084	920	90	20.86	21.29	21.07
	12	4	6,000	0.8769	0.8829	0.100	920	90	21.29	21.42	21.36
	13	4	10,000	0.8829	0.8829	0.00**	750	110	16.52	15.52	16.52
	14	4	20,000	0.8829	0.8829	0.00**	750	100	16.78	16.78	16.78
	15	4	10,000	0.8829	0.8829	0.00**	750	100	16.78	16.78	16.78

* During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

** A sudden shift from a large to small range of load variation has apparently stopped crack extension for the load cycles applied. It is discussed in the text on page

Table XVIII. Fatigue crack propagation test data of alloys after 20% deformation at 450°C (cont.)

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dn (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum Pmax	Minimum Pmin	Start	End	Average
S-22, cont.	16	4	10,000	0.8829	0.9168	0.339	1200	150	27.11	28.99	28.05
	17	4	10,000	0.9168	0.9616	0.448	350	200	28.99	34.83	31.91
	18	4	4,500	0.9616	1.1518	4.22	1460	200	38.17	57.70	47.94
	19	1	1,000	1.1518	1.2159	6.41	1200	150	48.08	54.16	51.12
S-23	2	4	20,000	0.7045	0.7976	0.466	1200	150	20.10	24.40	22.25
	3	4	3,000	0.7976	0.8148	0.573	1200	150	24.40	25.26	24.83
Alloy a ₁	4	4	10,000	0.8148	1.0108	1.960	1200	150	25.26	36.75	31.01
	5	2	2,000	1.0108	1.1300	5.960	1200	150	36.75	51.54	44.15
	6	1	500	1.1300	1.2268	19.36	1430	200	60.38	69.15	64.77
	7	1	100	1.2268	1.2565	29.7	1450	165	72.24	76.10	74.17
	8	1	100	1.2565	1.3222	65.7	1570 - 1510*	185	82.02	86.77	84.40
S-28	2	4	15,000	0.6820	0.8342	1.015	1890	200	29.14	39.14	34.14
	3	4	5,000	0.8342	0.9832	2.98	1900	250	39.14	52.28	45.71

* During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

Table XVIII. Fatigue crack propagation test data of alloys after 20% deformation at +50°C (cont.)

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dN (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum Pmax	Minimum Pmin	Start	End	Average
S-28, cont.	1	2	1,000	0.9832	1.1242	14.10	1800	250	49.11	66.89	58.00
Alloy a_2	5	1	50	1.1242	1.2003	152.2	2325	300	87.39	103.57	95.48

Table XIX. Fatigue Crack Propagation Test Data of Alloys after 20% Deformation at 250°C.

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate $\frac{da}{dn}$ (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum Pmax	Minimum Pmin	Start	End	Average
S-40	2	4	20,000	0.6836	0.7110	0.132	930	60	15.38	15.82	15.60
Alloy b ₁	3	4	5,000	0.7110	0.7870	1.520	1475	95	25.66	29.54	27.60
S-42	2	4	10,000	0.7100	0.7440	0.340	1050	135	18.88	19.80	19.34
	3	4	5,000	0.7440	0.8008	1.156	1260	150	21.70	24.48	23.09
Alloy b ₁	4	4	3,000	0.8008	0.8556	1.827	1350	200	25.32	28.31	26.81
	5	2	1,500	0.8556	0.9028	3.141	1605	300	32.13	35.18	33.66
	6	1	1,200	0.9028	1.0020	8.257	1880	410	39.63	48.16	43.90
	7	1	5,000	1.0020	1.1226	24.120	2125	475	54.06	60.18	57.12
	8	1	2,200	1.1226	1.2732	68.50	2310 - 2145*	580	63.10	90.51	76.81
	9	C.5	50	1.2732	1.4401	333.8	2020	500	88.20	115.00	101.60
S-44	2	4	20,000	0.7089	0.8278	0.595	1020	90	17.79	22.61	20.70
	3	4	20,000	0.8278	0.9942	0.832	885	90	22.20	30.77	26.49
Alloy b ₂	4	2	2,000	0.9942	1.0846	4.52	1210	190	34.37	42.55	38.46

*During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

Table XX. Fatigue Crack Propagation Test Data of Alloys After 80% Deformation at 250°C.

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dN ($\text{in.} \times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P_{max}	Minimum P_{min}	Start	End	Average
S-30 Alloy c_1	2	4	10,000	0.6734	0.6745	0.011	1050	120	15.96	16.19	16.08
	3	4	50,000	0.6745	0.7746	0.200	1060	130	16.19	19.54	17.87
	4	4	10,000	0.7746	0.8234	0.488	1120	130	22.86	25.21	24.04
	5	4	3,000	0.8234	0.8549	1.050	1360	150	28.04	29.80	28.92
	6	2	1,500	0.8549	0.8877	2.187	1595	200	34.34	36.47	35.41
	7	1	500	0.8377	0.9126	4.98	1700	200	41.17	39.22	40.20
	8	1	500	0.9126	0.9485	7.18	1880	230	45.29	48.55	46.92
	9	1	500	0.9485	1.0371	17.72	2040	240	52.96	65.23	59.10
	10	1	150	1.0371	1.0850	31.93	2100	300	52.96	72.80	62.88
	11	1	80	1.0850	1.1467	77.13	2260	300	79.27	89.52	84.40
	2	4	5,000	0.6933	0.6986	0.106	1500	240	19.70	19.96	19.83
S-35 Alloy c_2	3	4	2,000	0.6986	0.7624	0.315	1815	300	24.00	26.97	25.49
	4	2	750	0.7524	0.7730	1.413	2100 - 2330*	380	30.62	35.80	33.21
	5	-	500	0.7730	0.7854	2.48	2520	420	38.55	39.30	38.92
	6	1	500	0.7854	0.8019	3.3	2760	450	43.23	44.42	43.83

During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

Table XX. Fatigue Crack Propagation Test Data of Alloys After 80% Deformation at 250°C. (cont.)

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dN (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P _{max}	Minimum P _{min}	Start	End	Average
S-35	7	1	500	0.8019	0.8480	9.22	3200	600	50.00	55.11	52.55
	8	1	200	0.8480	0.8600	16.00	3480	600	61.04	64.52	62.78
	9	1	200	0.8800	0.9474	33.70	3950	720	72.36	83.02	77.69
	10	1	100	0.9474	1.0004	53.00	4050	860	81.99	90.93	86.46
	11	1	100	1.004	1.0984	93.00	4200	960	92.36	117.20	104.78
	12	0.5	30	1.0984	1.2118	378.00	4100 - 3550*	180	121.80	131.26	126.53

*During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

Table XXI. Fatigue Crack Propagation Test Data of Alloys After 80% Deformation at 450°C.

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dr (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P _{max}	Minimum P _{min}	Start	End	Average
S-4	2	6	3,000	0.630	0.718	2.93	2400	300	32.10	37.88	35.00
Alloy d ₁	3	6	2,000	0.718	0.809	4.55	2400	280	37.98	45.81	41.90
	2	6	10,000	0.681	0.907	2.23	1600	200	24.78	36.52	30.65
S-5	3	6	6,000	0.907	1.105	3.30	1200	150	27.85	41.85	34.85
Alloy d ₁											
S-6	2	6	8,000	0.725	0.792	0.835	2000	250	31.00	35.40	33.20
Alloy d ₁	3	6	16,500	0.792	0.907	0.698	1600	200	35.00	28.00	31.50
	4	6	6,500	0.907	1.081	2.680	1600	200	35.00	51.50	43.25
S-11	2	4	20,000	0.6932	0.6932	0.00	560	70	8.43	8.43	8.43
Alloy d ₂	3	4	20,000	0.6932	0.6932	0.00	600	110	8.43	8.43	8.43
	8	4	20,000	0.6932	0.6932	0.00	600	75	9.03	9.03	9.03
	9	4	50,000	0.6932	0.6932	0.00	625	95	9.12	9.12	9.12
	10	4	50,000	0.6932	0.6932	0.00	750	110	11.01	11.01	11.01
	11	4	50,000	0.6932	0.6932	0.00	900	130	13.25	13.25	13.25
	12	4	50,000	0.6932	0.6932	0.00	1050	150	15.49	15.49	15.49
	13	4	10,000	0.6932	0.6932	0.00	1170	160	17.39	17.39	17.39

Table XXI. Fatigue Crack Propagation Test Data of Alloys After 80% Deformation at 450°C (continued).

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dN (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P _{max}	Minimum P _{min}	Start	End	Average
S-12	15	4	3,000	0.6932	0.9412	8.467	3300	450	49.06	80.67	64.83
	16	1	1,000	0.9472	1.1739	226.50	3300	450	80.67	132.83	106.75
	2	4	10,000	0.7100	0.7416	0.316	1295	150	20.76	22.02	21.39
	3	4	10,000	0.7416	0.7942	0.526	1185	165	19.62	21.71	20.66
	4	4	10,000	0.7942	0.8584	0.642	1180	170	21.50	24.22	22.86
Alloy d ₂	5	4	10,000	0.8584	0.9272	0.688	1180	170	24.23	27.77	26.00
	6	4	5,000	0.9272	0.9723	0.902	1180	170	27.50	30.33	28.92
	7	2	2,500	0.9723	0.9917	1.253	1180	180	30.33	31.40	30.87
	8	2	2,500	0.9917	1.0116	1.327	1180	180	31.40	32.18	31.79
	9	2	1,000	1.0116	1.0252	1.360	1170	150	32.82	34.71	33.76
	10	2	1,000	1.0252	1.0519	2.670	1200	190	34.36	36.63	35.50
	11	1	1,000	1.0519	1.0796	2.770	1200	180	37.00	39.40	38.20
	12	1	1,000	1.0796	1.1152	3.560	1200	180	39.40	42.50	40.95
	13	1	1,000	1.1152	1.1594	4.420	1210	180	42.91	46.93	44.92
	14	1	1,000	1.1594	1.1979	3.850	1200	180	46.48	50.07	48.28
	15	1	1,000	1.1979	1.2461	4.820	1200	180	50.07	54.00	52.04
	16	1	1,000	1.2461	1.3088	6.270	1185	180	53.20	58.91	56.06

Table XXII. Fatigue Crack Propagation Test Data of Alloys After 80% Deformation at 450°C (continued).

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dN (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P_{max}	Minimum P_{min}	Start	End	Average
S-15	2	4	5,000	0.7481	0.7759	0.556	1220	120	21.12	22.24	21.68
	3	4	2,000	0.7759	0.7960	1.005	1420	140	25.82	27.04	26.43
	4	1	1,000	0.7560	0.8144	1.840	1730	80	34.94	36.18	35.56
	5	1	1,000	0.8144	0.8368	2.240	1925	150	38.92	40.51	39.71
	6	1	1,000	0.8368	0.9098	7.300	2130	300	41.77	48.04	44.91
	7	1	500	0.9098	0.9866	15.72	2300	370	50.67	59.10	54.88
	8	1	200	0.9866	1.0397	26.60	2640	660	60.63	69.02	64.83
	9	0.5	100	1.0397	1.1130	73.30	2950	660	79.83	110.0	94.91
	10	0.5	40	1.1130	1.2508	344.50	3090 - 2920*	750 - 700*	112.39	117.60	115.00
S-51	2	4	10,000	0.6890	0.7087	0.197	1200	150	18.31	19.04	18.67
	3	4	10,000	0.7087	0.7214	0.127	1230	150	19.58	19.83	19.70
Alloy d ₂	4	4	10,000	0.7214	0.7484	0.270	1500	200	23.87	25.29	24.58
	5	4	10,000	0.7484	0.7740	0.256	1480	200	25.29	26.21	25.75
	6	4	10,000	0.7740	0.7985	0.245	1520 - 1460*	200	27.03	26.78	26.90
	7	4	10,000	0.7585	0.8417	0.432	1660	200	31.03	34.01	32.52
	8	4	10,000	0.8417	0.8907	0.490	1660	200	34.01	37.66	25.84

*During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

Table XXI. Fatigue Crack Propagation Test Data of Alloys After 80% Deformation at 450°C (continued).

Specimen and alloy identification number	Run No.	Rate of Cycling f cps	No. of Incremental cycles ΔN	Crack Length, a (in)		Growth Rate da/dN (in. $\times 10^{-5}$)	Applied cyclic Load, P (lbs)		Stress Intensity Factor Range, ΔK		
				Start	End		Maximum P _{max}	Minimum P _{min}	Start	End	Average
	9	4	10,000	0.8907	0.9300	3.93	1680	150	39.47	40.53	41.00
	10	4	2,000	0.9300	1.0249	4.750	1730	200	42.53	54.53	48.53
	11	1	1,000	1.0249	1.1694	14.45	1730 - 1680*	200	54.53	69.83	62.18

* During this particular run, the load slowly and gradually changed. The values of the load at the beginning and at the end of the run are given.

FIGURE CAPTIONS

- Fig. 1. Low cycle fatigue specimen.
- Fig. 2. Tensile sheet specimen.
- Fig. 3. Room temperature fatigue testing jig.
- Fig. 4. 200°C fatigue testing jig.
- Fig. 5. Photograph of the complete MTS assembly for room temperature testing including the strain gage instrumentation.
- Fig. 6. Single edge notched specimen.
- Fig. 7. Schematic illustration of the "tensile permeameter" designed for measurement of the amount of martensite magnetically in flat specimens during tensile testing.
- Fig. 8. Schematic illustration of low cycle fatigue permeameter, designed for measuring the amount of martensite in the fracture area of low cycle fatigue specimen.
- Fig. 9. Typical stress-strain behavior upon strain cycling.
- Fig. 10. "Stable" hysteresis loop of a strain cycled specimen.
- Fig. 11. Schematic illustration of (a) cyclic strain hardening and (b) cyclic strain softening materials.
- Fig. 12. The three basic modes of crack surface displacements.
- Fig. 13. K calibration for SEN specimen.
- Fig. 14. True stress-strain curves for the alloys deformed at 450°C (Round specimens).
- Fig. 15. True stress-strain curves for the alloys deformed at 250°C (Round specimens).
- Fig. 16. Engineering stress-strain curve (continuous curve) and volume percent martensite-strain curve (broken curve) for the alloys a and d deformed at 450°C (sheet specimens).

- Fig. 17. Same as above figure for the alloys b, and c deformed at 250°C (sheet specimens).
- Fig. 18. Microstructures of (a) Alloy d₁, flat rolled 80% at 450°C, (b) Alloy c₁, flat rolled 80% at 250°C, (c) Alloy D, form rolled 80% at 450°C, (d) Alloy C, form rolled, 80% at 250°C.
- Fig. 19. Microstructure of (a) Alloy a₁, flat rolled 20% at 450°C (b) Alloy b₁, flat rolled 20% at 250°C.
- Fig. 20. Typical hysteresis loops for Alloy D (80% deformed at 450°C) with diametral strain ranges of (a) $\epsilon_t^d = 8.00\%$ (b) $\epsilon_t^d = 2.50\%$.
- Fig. 21. Plastic strain range vs cycles to failure for Alloy A at room temperature and 200°C.
- Fig. 22. Plastic strain range vs cycles to failure for Alloy B at room temperature and 200°C.
- Fig. 23. Plastic strain range vs cycles to failure for Alloy C at room temperature and 200°C.
- Fig. 24. Plastic strain range vs cycles to failure for Alloy D at room temperature and 200°C.
- Fig. 25. Plastic strain range vs cycles to failure for Alloy A with and without prestrain (room temperature). The 'x' mark is N = 0.25 point based on fatigue ductility.
- Fig. 26. Same as Fig. 25 for Alloy B.
- Fig. 27. Same as Fig. 25 for Alloy C.
- Fig. 28. Same as Fig. 25 for Alloy D.
- Fig. 29. Stress range versus strain cycles for Alloy A at room temperature. The number in the brackets is the plastic strain range and the outer number is total strain range.

Fig. 30. Same as Fig. 29 for Alloy B.

Fig. 31. Same as Fig. 29 for Alloy C.

Fig. 32. Same as Fig. 29 for Alloy D.

Fig. 33. Stress range versus strain cycles for Alloy A at 200°C.

The number in the brackets is the plastic strain range and the outer number is the total strain range.

Fig. 34. Same as 33 for Alloy B.

Fig. 35. Same as 33 for Alloy C.

Fig. 36. Same as 33 for Alloy D.

Fig. 37. Volume percent martensite in the fracture area vs cyclic plastic strain range for the low cycle fatigue specimens.

The number in the brackets are the cycles to failure.

Fig. 38. Fracture surface appearance of Alloy A, deformed 20% at 250°C after various cycles to fracture (a) 25, (b) 46, (c) 81, (d) 276, (e) 771, (f) 1460.

Fig. 39. Schematic illustration of typical fractures: (a) Monotonic fracture (cup and cone) (b) fracture at very low cycles, (c) fracture at (cup and cone) somewhat higher cycles.

Fig. 40a+b. Fractographs of Alloy D, 80% deformed at 450°C. Tensile fracture exhibiting a wide variety of dimple sizes.

Fig. 41. Fractograph of the fracture surface of Alloy D specimen (80% deformed at 450°C) after failure at 158 cycles showing uniformly elongated dimples.

Fig. 42. Fractographs of the fracture surface of Alloy D (80% deformed at 450°C) specimen after failure at 381 cycles showing "tire tracks".

- Fig. 43. Same as Fig. 42, failed after 873 cycles. Top half shows a feature similar to striations superimposed with dimples.
- Fig. 44. Fractographs of the Alloy A (deformed 20% at 450°C) specimen which failed after 365 cycles at 200°C. A large amount of plastic deformation, serpentine glide and considerable degree of stretching can be seen.
- Fig. 45. Scanning electrons micrographs showing fracture surfaces of the Alloy D specimen (deformed 80% at 450°C) which failed after 22 cycles. A wide variety of dimple sizes can be seen.
- Fig. 46. Same as Fig. 45, but failed after 98 cycles. Rubbing marks resembling striations can be seen in (c) and (d).
- Fig. 47. Same as Fig. 45, but failed after 1514 cycles. Well-developed striations can be seen in (c), (e) and (f).
- Fig. 48. Scanning electron micrographs showing predominance of ductile features in the fracture surfaces of the Alloy A specimen (deformed 20% at 450°C) which failed after 32 cycles.
- Fig. 49. Similar as Fig. 48, but failed after 64 cycles.
- Fig. 50. Similar as Fig. 48, but failed after 124 cycles.
- Fig. 51. Similar to Fig. 48, but failed after 1356 cycles.
- Fig. 52. da/dn vs ΔK curve for the alloys deformed 20% at 450°C. Solid circles indicate microscopic crack propagation rates and the band shows the maximum possible scatter in ΔK values.
- Fig. 53. da/dn vs ΔK curve for the alloys deformed 20% at 250°C.
- Fig. 54. da/dn vs ΔK curve for the alloys deformed 80% at 250°C.
- Fig. 55. da/dn vs ΔK curve for the alloys deformed 80% at 450°C. Solid circles indicate microscopic crack propagation rates and the band shows the maximum possible scatter in ΔK values.

- Fig. 56. The da/dn vs ΔK trend for all the four processing treatments.
- Fig. 57. (a) Fatigue crack-tip in the SEN specimens for the alloy deformed 80% at 450°C. (b) Crack path in the SEN specimen for the same alloys. Plastic zone surrounding the crack path can be distinctly seen.
- Fig. 58. Fatigue crack path in the SEN path for the alloy deformed a-b 20% at 450°C. The arrows in (a) indicate branching cracks.
- Fig. 59. Fractographs from the SEN specimen of the alloy deformed 20% a-b at 450°C showing the marks and cleavage features. Some artifacts also can be seen in the fractograph.
- Fig. 61. Fractographs from the SEN specimen of the alloy deformed 80% at 450°C. (a) and (b) show striations while (c) shows elongated dimples.
- Fig. 62. Fractographs from the SEN specimen of the alloy deformed 80% at 450°C. (a) shows brittle striations while (b) shows flat fracture.
- Fig. 63. Room temperature cyclic and monotonic stress-strain curves for the alloys deformed at 450°C.
- Fig. 64. Same as Fig. 63 for the alloys deformed at 250°C.
- Fig. 65. 200°C cyclic and monotonic stress-strain curves for the alloys deformed at 450°C.
- Fig. 66. Same as Fig. 65 for the alloys deformed at 250°C.
- Fig. 67. Plastic strain range vs cycles to failure for alloys deformed 80%.
- Fig. 68. Plastic strain range vs cycles to failure for alloys deformed at 250°C.

- Fig. 69. Plastic strain range vs cycles to failure for alloys deformed at 450°C.
- Fig. 70. Tensile ductility vs fatigue ductility for various alloys.
- Fig. 71. Comparison of crack-propagation rates of 250 ksi ultimate TRIP steels (80% deformed at 250°C) with other high strength alloys in the same tensile strength range.
- Fig. 72. Comparison of crack-propagation rates of 185 ksi ultimate TRIP steels (20% deformed steels) with AM 350 CRT steel of 160-185 ksi tensile strength.
- Fig. 73. Percent martensite in fracture area of low cycle fatigue specimen vs number of cycles to failure for various cyclic plastic strain ranges.

XBL 704-812

Figure 1

GAGE LENGTH: 1"
THICKNESS: 0.05"

TENSILE SPECIMEN

XBL 705-911

Figure 2

XBL 698-1197

Figure 3

XBL 705-1038

Figure 4

XBB697-4447

Figure 5

XBL 705-944

Figure 6

XBL 706-1295

Figure 7

Figure 8

XBL 706-1287

XBL 705-910

Figure 9

XBL 704-815

Figure 10

Figure 11

XBL 706-1294

XBL 704-825

Figure 12

XBL 706-1108

Figure 13

XBL 706-1296

Figure 14

XDL 706-1298

Figure 15

XBL 706-1297

Figure 16

XBL 706-1299

Figure 17

(a)

(b)

(c)

(d)

XBB706-2808

Figure 18

(a)

(b)

XBB706-2809

Figure 19

(a)

(b)

XBL 706-1288

Figure 20

Figure 21

XBL 704-871

Figure 22

XBL 704-872

Figure 23

XBL 704-876

Figure 24

XBL 704-874

Figure 25

XBL 704-875

Figure 26

XBL 704-877

Figure 27

XBL 704-870

Figure 28

XBL 704-869

Figure 30

XBL 704-806

Figure 29

XBL 704-808

Figure 32

XBL 704 809

Figure 34

XBL 705-941

Figure 34

XBL 706-1283

Figure 35

NBL 705-942

Figure 36

XBL 705-943

Figure 37

(a)

(b)

(c)

(d)

(e)

(f)

1 cm

XBB706-2792

Figure 38

XBL 706-1223

Figure 39

(a)

(b)

XBB706-2740

Figure 40

Figure 41

XBB706-2737

(a)

XBB706-2736

(b)

Figure 42

(c)

(d)

XBB706-2741

Figure 42 (continued)

Figure 43

XBB706-2735

(a)

(b)

XBB706-2739

Figure 44

(a)

(b)

(c)

(d)

XBB707-2931

Figure 45

(a)

(b)

(c)

(d)

XBB704-2015

Figure 46

(a)

(b)

XBB704-2010

Figure 47

(c)

(d)

(e)

(f)

XBB/04-2012

Figure 47 (continued)

(a)

(b)

(c)

(d)

XBB704-2011

Figure 48

(a)

(b)

XBB704-2007

Figure 49

(a)

(b)

XBB706-2505

Figure 50

(a)

(b)

(c)

(d)

XBB704-2014

Figure 51

Figure 52

XBL 706-1289

Figure 53

XBL 706-1285

Figure 54

XBL 706-1286

Figure 55

XBL 706-1291

Figure 56

XBL 706-1284

(a)

(b)

XBB706-2791

Figure 57

(a)

(b)

XBB706-2790

Figure 58

(a)

(b)

XBB706-2918

Figure 59

XBB706-2917

Figure 60

(a)

(b)

(c)

XBB706-2915

Figure 61

(a)

(b)

XBB706-2916

Figure 62

Figure 63

XBL 707-1324

Figure 64

Figure 65

XBL 707-1325

Figure 66

XBL 707-1326

Figure 67

XBL 704-819

Figure 68

XBL 704-817

Figure 69

XBL 704-818

Figure 70

XBL 707-1327

Figure 71

XBL 706-1290

Figure 72

XBL 706-1292

XBL 706-1300

Figure 73

Figure 1-A

XBL 706-1107

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