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THE EFFECT OF ENVIRONMENT ON FATIGUE CRACK
PROPAGATION IN 4340 STEEL

by

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ABSTRACT

Environmental fatigue crack growth rate studies were conducted with a 4340 steel heat treated to several different yield strengths in hydrogen, argon, room air, and a .4% by volume oxygen-hydrogen atmosphere at frequencies of 5, .5 and .05 cycles per second. The fracture mechanics approach was used.

A decrease in the severity of the hydrogen induced cracking with a decrease in the materials yield strength was observed. This phenomena was explained in terms of Troiano's maximum stress triaxiality theory in which the diffusion of atomic hydrogen to the point of maximum stress triaxiality is the rate controlling factor.

The beneficial effect of a small oxygen addition to the aggressive hydrogen atmosphere was demonstrated. It was proposed that this was due to the preferential adsorption of an oxide film on the crack tip surface which cyclically ruptures and reforms, hence preventing a damaging amount of hydrogen from being adsorbed and diffusing to the point of maximum stress triaxiality.

CHAPTER I. LITERATURE REVIEW

A. Introduction

The detrimental effect of certain environments on the load bearing performance of numerous engineering alloys has been well documented. The interaction of hydrogen, water vapor and salt water environments with an applied stress has induced premature failures in a wide variety of materials, including high strength steels and numerous magnesium, titanium, aluminum and nickel base alloys⁽¹⁻⁶⁾. The most catastrophic and most frequently investigated environmentally induced failures occur in high strength steel alloys.

The stress-environment interaction in high strength steels is of great interest. These alloys are used in applications where yield strength-specific weight ratios in excess of 700,000 inches are required. Current processing and fabrication techniques cannot produce a part completely free of microscopic flaws and cracks. Numerous investigators have demonstrated that rapid crack initiation and propagation may occur at these pre-existing flaws at unusually low stress levels or, in the case of fatigue applications, after an unusually low number of cycles when the material is exposed to an aggressive environment such as hydrogen or water vapor⁽⁷⁻¹⁷⁾. Rapid and in most cases unexpected brittle fracture will occur when the crack has grown to a certain critical size, characteristic of the

fracture toughness of the material.

B. Fracture Mechanics

All environmentally induced failure theory calls for a stress-environment interaction in the area of the fissure or microcrack tip. The most informative method of studying this phenomenon is the fracture mechanics approach which has its origins in the Griffith crack surface energy theory⁽¹⁸⁾.

Griffith laid the cornerstone for fracture mechanics by proposing that an existing crack will propagate in a catastrophic manner if the increase in the surface energy of the crack is exceeded by the available elastic strain energy release rate. For extremely brittle materials the Griffith theory yielded accurate predictions of conditions for unstable crack propagation. However, the theory could not be applied to semi-brittle materials; this led Orowan⁽¹⁹⁾ and Irwin⁽²⁰⁾ to modify the surface energy term by adding an expression to it which accounted for the work required to plastically deform the material at the crack tip.

Fracture mechanics was developed as an alternative to the Griffith energy approach. It is based on the application of a reasonably accurate elastic stress analysis to the crack tip region. A stress analysis based on the theory of elasticity is used with certain modifications to take into account the crack tip plasticity. The rudiments of fracture mechanics have been rigorously developed and are reviewed in great detail by numerous theorists⁽²¹⁻²⁶⁾. Only

the salient points will be discussed here.

The key concept in fracture mechanics is that fracture will occur when a critical stress field intensity, represented by K_c or K_{Ic} , the stress field intensity factor, is reached at the crack tip. A linear elasticity analysis of the stress distribution at the crack tip is used to derive the stress intensity factor. The three basic modes of crack deformation and the coordinate system for the stress distribution ahead of a sharp crack are shown in Figure 1. The most commonly occurring engineering material fractures are due to cracks opening directly as in Mode I. The simplest mathematical representation for this mode is developed with the help of Westergaard's stress field equations⁽²⁷⁾ for cracks in which any root radius or plastic zone effects at the crack tip are neglected. A summary of the Mode I stress fields is given in Table I. In the direction parallel to the crack edge the stress, σ_z , would equal $\nu(\sigma_x + \sigma_y)$ for the plane strain situation or zero for the plane stress state.

It is evident from Table I that identical local stress fields are obtained for identical stress intensity factor values and that the stress intensity factor is a function of applied stress and crack geometry. Therefore the stress conditions determining when the specimen will fracture are described by the stress intensity factor, K , and the degree of plane strain, i.e., the elastic confinement. A deeply embedded crack in a thick specimen represents a state of plane strain; the specimen's thickness provides a con-

Figure 1: Fracture Mechanics Nomenclature

- a. Coordinate system and stress distribution for the crack tip stress field
- b. The three basic modes of crack surface displacement

TABLE I

Crack Tip Stress Fields for Mode I Deformation
in an Isotropic Elastic Body

$$\sigma_x = \frac{K_I}{(2\pi r)^{1/2}} \cos \frac{\theta}{2} [1 - \sin \frac{\theta}{2} \sin \frac{3\theta}{2}]$$

$$\sigma_y = \frac{K_I}{(2\pi r)^{1/2}} \cos \frac{\theta}{2} [1 + \sin \frac{\theta}{2} \sin \frac{3\theta}{2}]$$

$$\tau_{xy} = \frac{K_I}{(2\pi r)^{1/2}} \sin \frac{\theta}{2} \cos \frac{\theta}{2} \cos \frac{3\theta}{2}$$

$$\sigma_z = \nu(\sigma_x + \sigma_y)$$

$$\tau_{xy} = \tau_{yz} = 0$$

where:

K_I = stress intensity factor for opening Mode I

ν = Poisson's ratio

and other terms as defined in Figure 1.

straint to deformation in the z-direction. A crack extending through the plate thickness with a significant plastic zone causing thickness direction stress relaxation represents a situation of plane stress.

Johnson and Paris⁽²⁸⁾ have provided an extensive discussion of the applicability of the fracture mechanics approach to both static and fatigue loading situations. The question of characterizing sub-critical crack growth by either the net section stress or the stress intensity factor was debated. Citing the experimental data of various investigators, it was concluded that a constant crack growth rate is a unique function of a constant stress intensity factor or in the case of fatigue testing, a constant stress intensity factor range. Hence, different test specimens, flaw geometries and loading configurations should all yield the same crack growth rates when tested at the same stress intensity level under the same environmental conditions.

The fast fracture problem is most critical in specimens where the plane strain situation prevails. The energy consuming plastic deformation at the crack tip is minimized so that crack propagation will occur at the lowest possible stress intensity factor. An excess amount of crack tip plastic deformation will cause a shift from a plane strain state to a plane stress state. Thus, it was necessary to make some mathematical estimate of the area of plastic deformation, i.e., the plastic zone size, in order to eliminate an experimental dependence of K_{IC} on the initial crack size,

and to determine a critical plastic zone size where K_{Ic} values become invalid parameters for predicting rapid fracture⁽²¹⁾.

An estimate of the plastic zone size, r_y , for static testing can be obtained from the stress equations in Table I by setting σ_y equal to $\sigma_{y.s.}$, the yield strength of the material. The plastic zone size, r_y , represents the distance r from the origin in Figure 1 to the elastic-plastic boundary directly ahead of the crack tip and is given by:

$$r_y = \frac{K_I^2 (24)}{2\sigma_{y.s.}^2} \quad (1)$$

When errors in K_{Ic} exceeding 10% result from the plastic deformation, the plastic zone size must be added to the actual crack length, a , and K_{Ic} must be recalculated using the newly found apparent crack length.

C. Fatigue Crack Propagation and Fracture Mechanics

In fatigue testing localized plastic deformation at the crack tip is of prime importance in the crack propagation process. A schematic diagram of the development of the crack tip plastic zone is shown in Figure 2 for an ideal elastic material⁽²⁹⁾. The plastic zone size is directly related to the stress intensity factor excursion, h_k . At the first peak in the K - time history, Point ①, the plastic zone size is estimated from the previously mentioned in Equation (1). A compressive zone is then formed during the excursion, h_k , of the stress intensity factor to Point ②.

Figure 2: Development of Crack Tip Plastic Zone in Elastic Plastic Material

The body is attempting to elastically unload in going from Point ① to Point ② by subtracting the reduction in elastic stress,

$$\sigma_y = \frac{K_{②} - K_{①}}{\sqrt{2r}} \quad (2)$$

where: $h_k = K_{②} - K_{①}$

This subtracted stress causes some of the material in tension at σ_y to exceed $\sigma_{y.s.}$ in compression due to the singularity in the denominator of Equation (2). By substituting $2\sigma_{y.s.}$ into Equation (2) one obtains the extent of this yielding to be

$$r_p = \frac{h_k^2}{8\sigma_{y.s.}^2} \quad (3)$$

When the K - Excursion is increased back to Point ①, a new tensile plastic zone forms of an extent predicted by Equation (3).

This explains the importance of the stress intensity factor excursion, h_k , in fatigue testing. There is no mention of the frequency of load application or the mean load in the above discussion. If plasticity in the crack tip area is the important variable then the frequency and the mean load should not strongly influence crack growth rates in a non-aggressive environment. This view is consistent with experimental results (30,31,32)

D. Environmentally Induced Failure Theories

It has been demonstrated that there are numerous simi-

larities between environmentally induced failures caused by a water or water vapor atmosphere and those observed in a hydrogen atmosphere⁽³³⁾. These similarities include nearly identical crack growth rates, reversible incubation periods for crack initiation, a definite stress intensity value below which delayed failure is not observed, and an increase in the brittle behavior at higher strength levels.

Thus, it has been assumed that hydrogen is the embrittling agent; the water and water vapor supply the hydrogen to the material's surface, where it is adsorbed. Several theories have been proposed which attempt to explain the numerous observed hydrogen induced failures.

Petch and Stables⁽³⁴⁾ theorized that hydrogen atoms were adsorbed on the surface of a pre-existing microcrack. These adsorbed hydrogen atoms were thought to lower the surface energy and hence the stress required to cause the microcrack to extend. The extension of the crack was assumed to occur in distinct steps governed by the rate at which the hydrogen atoms arrived at the edge of the crack. As pointed out by Johnson⁽³³⁾, one of the shortcomings of this theory is the lack of clarity as to the mechanism of formation of the clean crack prior to adsorption.

Another explanation for observed environmentally induced failures was proposed by Zappfe and Sims⁽³⁵⁾ and elaborated upon by Tetelman and Robertson^(36,37). This theory proposes that atomic hydrogen precipitates in numerous voids and cracks in the lattice and in the course of deformation

recombines to form molecular hydrogen. This molecular hydrogen supposedly builds up a significant internal stress which in turn lowers the external applied stress necessary to cause crack propagation and brittle fracture. Sawicki⁽³⁸⁾ has noted several of the questionable aspects of this theory, the most noteworthy being various examples of hydrogen induced failures of high strength steels at atmospheric pressures or less.

A third school of thought, that of Troiano, et al^(39,40), hypothesized that hydrogen diffuses to points within the lattice where a state of maximum triaxial stress exists. The concentration of hydrogen at these points acts to lower the theoretical cohesive strength of the material. This theory has the advantage of explaining the disappearance of the hydrogen induced failures at very high and very low temperatures. However, a demonstration of the exact mechanism by which the theoretical strength of the steel is lowered after the hydrogen diffuses to the points of maximum triaxiality has yet to be performed experimentally.

E. The Effect of Oxygen on Environmental Fatigue Crack Propagation

Although different explanations as to the actual mechanism by which external hydrogen embrittles steel are presently being debated, it is generally agreed that adsorption of atomic hydrogen on the surface of the crack tip is necessary. This suggests that the embrittling process might be prevented by the preferential adsorption of some other

element on the crack tip surface, e.g., oxygen, hence eliminating the endangering hydrogen adsorption.

Hancock and Johnson⁽⁴¹⁾ investigated the effects of small oxygen additions on subcritical crack growth under static loading conditions with H-11 steel in environments containing varying proportions of argon, nitrogen, water vapor and hydrogen. In all cases the addition of as little as .6% by volume of oxygen caused the crack growth to terminate. The effect of the oxygen additions was quite rapid and the cracks could be restarted only when the oxygen was removed from the external environment. These results were explained in terms of preferential oxygen adsorption at the crack tip and the subsequent formation of an oxide barrier.

The effect of small oxygen additions to a hydrogen environment was also demonstrated by Sawicki⁽³⁸⁾ using H-11 Steel by doing static constant stress intensity factor testing. It was shown that as little as a .4% by volume oxygen addition to a pure hydrogen environment was sufficient to prevent hydrogen induced crack growth between -28°C and 100°C. This was thought to be caused by the formation of a preferentially chemisorbed oxide layer at the crack tip. Only when the material was stressed enough to rupture this oxide film could the hydrogen enter the material.

These observations indicate that perhaps below a certain stress intensity factor excursion range hydrogen induced fatigue cracking might be prevented by small oxygen

additions to the aggressive atmosphere. Li, Talda and Wei⁽⁴²⁾ examined this possibility by fatigue testing a .45C - Ni - Cr - Mo high strength steel at a frequency of 125 cycles per second in several different environments. Fatigue crack growth rates were nearly identical for the humidified air and the humidified argon environments. There was no detectable reduction in the crack growth rate which might have been expected due to the oxygen in the air.

Li, et al, noted that the crack growth inhibiting effect of the oxygen observed by Hancock and Johnson may have been due to its particular concentration in the environment and that it may vary for different types of steels.

Wei, Talda and Li⁽⁴³⁾ conducted further tests on a variety of ultrahigh strength steels in different controlled environments at a frequency of 125 cycles per second. Carman and Katlin⁽⁴⁴⁾ had previously reported fatigue crack propagation data for 1/4 inch thick 18Ni(250) and 18Ni(300) maraging steels tested in "room" air at a frequency of 4 cycles per second. Wei, et al's test results with the same maraging steels of the same thickness fatigued in the humidified environment were compared with the data of Carman and Katlin because Li, et al⁽⁴²⁾ had shown that these results should be comparable to data obtained in room air. It can be seen in Figures 3 and 4 that Wei, et al's results merge well with those of Carman and Katlin at the higher rates of crack growth. From this comparison it was concluded that any variation in the cycling frequency be-

Figure 3: Comparison of Crack Propagation Rates from Carman and Katlin with Results of Wei, et.al., on 18 Ni (250) Maraging Steels⁴³

Figure 4: Comparison of Crack Propagation Rates from Carman and Katlin with Results of Wei, et.al., on 18 Ni (300) Maraging Steels⁴³

tween 4 and 125 cycles per second would have little effect on the crack propagation rate. It was also observed that there was no discernible oxygen effect within this frequency range.

An interesting and fairly extensive study of the influence of humidified and dry argon atmospheres on fatigue crack propagation rates in H11 steel was made by Babcock⁽⁴⁵⁾. Data was accumulated at .007, .7, 7 and 30 cycles per second in order to observe any frequency effects which might have been present. Crack growth rates were found to be approximately an order of magnitude greater in the humidified argon environment for a given stress intensity factor excursion. The crack growth rates were also demonstrated to be frequency dependent as shown in Figure 5; the higher hydrogen induced crack growth rates were seen at the lower frequencies. Babcock explained his results in terms of Troiano's hydrogen embrittlement theory⁽³⁹⁾ which requires a critical combination of triaxial stress and hydrogen concentration in the region of maximum stress triaxiality. At lower frequencies more time per cycle is available for the hydrogen to diffuse to the region of maximum stress triaxiality and to hence enhance the growth process. It was also noted that the crack growth rates in the humidified argon and the dry argon atmospheres were approximately the same at the higher test frequencies.

Babcock also speculated on the possible protective influence of a hydrogen-oxygen atmosphere on the fatigue crack

Figure 5: Frequency Dependence of Crack Growth Rates at $\Delta K = 14 \text{ ksi } \sqrt{\text{in}}$ in H-11 Steel³⁹

growth behavior of a high strength steel. It was suggested that a small oxygen addition to the hydrogen environment would prevent fatigue crack propagation below a certain stress intensity factor level. At higher frequencies the protective effect of the oxygen would probably decrease; the oxide film would have less time per cycle to form allowing possible ruptures in it to occur which would permit the hydrogen adsorption and subsequent diffusion to take place.

The primary purpose of this present research is to accurately determine fatigue crack growth rates in both hydrogen and hydrogen-oxygen environments over a range of frequencies in order to observe and characterize any possible beneficial effects that a small oxygen addition may have.

F. Fatigue Crack Propagation Laws

A functional relationship between the crack growth rate da/dN , and the stress intensity factor excursion, ΔK , would be of great use in the problem of predicting the overall fatigue life of a particular structure. Such a relationship would indicate that fatigue crack growth rates are independent of specimen geometry and loading configuration. Hence, laboratory data obtained for constant ΔK testing could be useful in practical structure design.

Paris⁽⁴⁶⁾ suggested a power law relationship of this sort for sinusoidal loading,

$$\frac{da}{dN} = A(\Delta K)^n \quad (4)$$

where: a = crack length, inches

N = number of cycles

ΔK = stress intensity factor range, ksi $\sqrt{\text{in.}}$

A and n are constants for a given material

The exponent n is easily evaluated from the slope of a log-log plot of test data. Paris found n to be equal to 4 for test data gathered from a wide variety of materials.

A further study of the dependence of the fatigue crack growth rate on the stress intensity factor range was done by Miller⁽⁴⁷⁾ on 4340, H-11 and an 18Ni maraging steel, all heat treated to provide a wide range of mechanical properties. It was shown that n was not an invariant integer but ranged from 2.53 to 6.73 for these materials which were tested in "room" air. Also demonstrated was the fact that the fatigue crack resistance of a material is exclusively determined from the parameter A in Equation (4) because the exponent n is itself inversely related to the fracture toughness of the material.

In a more recent study the fatigue crack propagation characteristics of four high strength steels - HY-80, HY-130, 10Ni-Cr-Mo-Co and 12Ni-5Cr-3Mo - were investigated by Barsom, Imhof and Rolfe⁽⁴⁸⁾. A conservative crack propagation rate was determined to be

$$\frac{da}{dN} = .66 \times 10^{-8} (\Delta K)^{2.25} \quad (5)$$

for the high yield strength steels tested. Testing was done

in a room air environment at the ambient temperature. Comparison was made with existing fatigue crack growth data for numerous other high strength steels and with one exception Equation (5) accurately represented the fatigue crack behavior.

An interesting and potentially useful approach to the prediction of environmental fatigue crack growth rates in high strength steels was recently proposed by Wei and Landes⁽⁴⁹⁾. It was suggested that the mechanisms for hydrogen induced subcritical crack growth in high strength steels may be the same in both the sustained loading and the fatigue loading cases. This suggestion is supported by experimental fatigue data supplied by Li, Talda and Wei⁽⁴²⁾, Wei, Talda and Li⁽⁴³⁾ and Spitzig, Talda and Wei⁽⁵⁰⁾ in which all of the effects of water vapor, hydrogen and oxygen on sustained load subcritical crack growth were reproduced and observed. It was also emphasized that the fatigue crack growth rate versus K_{max} curve was nearly identical in shape to the growth rate versus K curve obtained from sustained load testing. The frequency dependence of the hydrogen induced fatigue crack growth rates was also cited.

Wei and Landes proposed that environmentally induced fatigue crack growth in high strength steels can be described by two independent components - a mechanical one and an environmental one. The overall rate of fatigue crack growth is taken to be equal to the algebraic sum of these two components. The mechanical component is the fatigue

crack growth rate in the absence of an aggressive environment and is experimentally determined by testing in an inert environment or a vacuum. The environmental component is computed from crack growth data obtained in the appropriate environment during sustained load testing and the appropriate fatigue load stress intensity factor profile.

Formally:

$$\left(\frac{\Delta a}{\Delta N}\right)_c = \left(\frac{\Delta a}{\Delta N}\right)_r + \int_{\tau} \frac{da}{dt} [K(t)] dt \quad (6)$$

where: $\left(\frac{\Delta a}{\Delta N}\right)_c$ = fatigue crack growth rate in the aggressive environment

$\left(\frac{\Delta a}{\Delta N}\right)_r$ = fatigue crack growth rate in an inert reference environment

$\left(\frac{da}{dt}\right)$ = sustained testing crack growth rate in the aggressive environment, and

$K(t)$ = a function representing the stress intensity factor profile incorporating the appropriate frequency and mean load

This method of analysis is schematically illustrated in Figure 6. The integration incorporates the effects of frequency and loading variables through $K(t)$ and is taken over one fatigue cycle.

Data from other experimentalists is analyzed in this fashion. It was found that this superposition method accurately predicts the frequency dependent environmentally induced fatigue crack growth rate curves. Wei and Landes

Figure 6: Schematic Diagram Illustrating the Suggested Method of Analysis: (a) stress intensity spectrum in fatigue (top left); (b) rate of crack growth under sustained load in an aggressive environment (top right); (c) environmental contribution to crack growth in fatigue (bottom left); (d) integrated effects on environment and K_{max} on fatigue crack growth rate (bottom right⁵⁰)

conclude that a simple correlation of fatigue crack growth rate with ΔK by means of an invariant power law would not be accurate due to environmental effects. It is argued that the exponent in Paris' power law is test frequency and load profile dependent and must be empirically determined in each individual case.

Thus, a secondary purpose of the present research study is to examine the fatigue crack growth data obtained in hydrogen and hydrogen-oxygen environments and to try and characterize it in light of the previously discussed crack propagation laws.

CHAPTER II. EXPERIMENTAL

A. Material

These experiments were performed on electric furnace 4340 aircraft quality steel supplied by Benedict-Miller, Inc. 4340 steel was selected for this research because of the relative ease of heat treating it to a wide range of yield strengths and because numerous other investigators in the environmental fatigue crack propagation area have used it, enabling comparisons with the present data to be readily made. The composition of the 4340 steel was: C, 0.40%; Mn, 0.65%; Si, 0.20%; Cu, 0.14%; Ni, 1.85%; Cr, 0.82%; Mo, 0.22%.

The specimens were divided into three different groups on the basis of heat treatment. They were heat treated to hardnesses of $50R_C$, $43R_C$ and $37R_C$ which produced yield strengths of approximately 215,000, 180,000 and 150,000 pounds per square inch, respectively. All of the specimens were austenitized at 1550°F for two hours after which they were oil quenched. The desired hardnesses were obtained by tempering at 600°F, 875°F and 1050°F for two hours.

B. Equipment and Procedure

The test specimens were of the compact crackline loaded type as shown in Figure 7. All were .1875 inches thick. By selecting this particular thickness the stress state in the material was varied from that of plane strain

Figure 7: Test Specimen Configuration

in the hardest specimens to that of plane stress in the softest ones, according to current ASTM criteria⁽⁵¹⁾.

The expression for the stress intensity factor excursion, ΔK , for the compact crackline loaded plate is given to be

$$\Delta K = \frac{\Delta P Y a^{\frac{1}{2}}}{B W}$$

where: ΔP

a = crack length

B = thickness

W = width

and Y is the result of a mathematical stress analysis⁽⁵¹⁾ for this particular specimen design and loading configuration and is shown in Figure 8. No plastic zone size correction was added to the measured crack length, a , because in all cases the error in ΔK due to plasticity effects was calculated and found to be less than 10%, and was hence not considered to be a serious problem.

The crack length was measured with a Model 632.01 Clip-on displacement gage supplied by the MTS Systems Co. It consists of a strain-gage temperature-compensated four-arm bridge extensometer which is attached to two cantilever beam-like arms. These arms are held between knife edges which are bolted to the specimen's side on both sides of the crack starting notch. The displacement in inches between the ends of the arms is derived by doing a calibration from the bending strain in the arms.

Figure 8: K-Calibrations for Compact Crackline Loaded Specimens⁵²

Crack length calibration was done by measuring the compliance, e.g., the displacement in inches of the cantilever arms divided by the applied load, for a known crack length. This was accomplished by fatiguing a crack of the desired length in the specimen and accurately measuring it with a jeweler's microscope; the corresponding compliance was then measured. The calibration curve obtained in this manner is plotted in Figure 9. It was estimated that the error in the compliance measurements, due for the most part to a zero drift in the load cell was $\pm .013 \times 10^{-5}$ in/lb. The corresponding error in the crack length measurements was computed from this estimate and varied according to the crack length and the length of crack growth increment being measured.

These tests were conducted in three different atmospheres - hydrogen, argon and a .4% by volume oxygen-hydrogen mixture. Several tests were run in room air on the intermediate yield strength specimens. Ultrahigh purity gases were used. The desired environment was circulated around the crack tip inside an aluminum environment chamber which was clamped to the specimen as shown in Figure 10. Gas leakage was prevented by a silicone-rubber sealant applied between the chamber and the specimen.

A 50,000 lb. closed loop electrohydraulic testing machine made by the MTS Systems Co. was used to load the specimens. Alignment was achieved with a set of pin grips held in a pair of self-aligning hydraulic wedge grips. All

Figure 9: Compliance vs Crack Length Calibration Curve

Figure 10: Environmental Chamber, Compliance Gage and Specimen Holders as Used During Testing

fatigue cycling was done over the load range zero to maximum load with a sinusoidal loading form. Precracking was done by cycling in room air at 5 cps. at a maximum ΔK of 18 ksi $\sqrt{\text{in}}$. The crack was then sharpened by further cycling in a hydrogen environment at a stress intensity level below that of any at which the specimen would be tested. This hopefully avoided any plastic deformation at the crack tip which might impede the progress of the crack propagation at the lower stress intensity levels.

After the precracking the testing was started. One specimen was used for one and only one combination of environment and test frequency. ΔK was varied by increasing increments with crack length measurements being made at an approximately constant stress intensity level. Cycling was never done below a stress intensity level which had been previously applied to the same specimen. The tests at the desired ΔK levels were terminated either after two hours of testing or when significant crack growth occurred. Testing was done at .05, .5 and 5 cps. in each environment.

Conversion of the yield strengths to metric units can be made by use of the following equality: $1 \text{ pound} / \text{in}^2 = .00072 \text{ kg/mm}^2$.

CHAPTER III. RESULTS AND DISCUSSION

A. Fatigue Crack Growth in Pure Hydrogen

The damaging effect of hydrogen on fatigue crack growth rates at all three strength levels is shown in Figures 11, 12 and 13. These fatigue crack growth rates were obtained under zero-tension loading at frequencies of .05, .5 and 5 cycles per second.

The 150,000 and 215,000 psi yield strength specimens were tested in pure hydrogen, argon and a .4% by volume oxygen-hydrogen mixture, while the 180,000 psi strength specimens were tested in pure hydrogen, room air and the same oxygen-hydrogen mixture.

Considering the argon to be an inert non-aggressive reference environment, the deliterious effect of the hydrogen on the fatigue crack growth behavior of 4340 steel can be readily seen. The damaging effect of the hydrogen is greatest in the highest yield strength specimens where the crack growth rates in the hydrogen atmosphere are two orders of magnitude greater than in the inert argon environment. The corresponding crack growth rate differences for the two lower strength levels is slightly less and one order of magnitude.

The influence of the cycling frequency on the fatigue crack growth behavior in hydrogen can be discerned from Figures 11, 12 and 13 and from the plot of da/dN vs $\sigma_{y.s.}$

Figure 11: da/dN vs ΔK for the 150,000 psi Yield Strength 4340 Steel

Figure 12: da/dN vs ΔK for the 180,000 psi Yield Strength 4340 Steel

Figure 13: da/dN vs ΔK for the 215,000 psi Yield Strength 4340 Steel

at constant ΔK in Figure 14. The frequency dependence of the crack growth rate which was reported by Babcock⁽⁴⁵⁾ was observed in the higher yield strength material, being most pronounced in the 215,000 psi yield strength specimens. This frequency dependence then diminished as the yield strength decreased until finally it disappeared in the 150,000 psi yield strength material. In all cases where a frequency dependence was observed the fastest crack growth rates were found at the lowest frequencies.

Babcock, in his work with H-11 steel, explained his observed frequency dependence of the crack growth rate in terms of Troiano's triaxial stress theory. Troiano's theory proposes that hydrogen is adsorbed at the crack tip and diffuses to the point of maximum stress triaxiality within the lattice where it subsequently lowers the theoretical cohesive strength of the material. A critical combination of hydrogen concentration and maximum stress triaxiality must be attained.

A qualitative discussion of the stress state ahead of the crack tip is given by Johnson⁽⁵²⁾. At the notch surface the stress state is biaxial; the longitudinal and tangential stresses are at a maximum while the radial stress is zero. However, the radial stress has a finite value at all other points along the notch cross-section creating a state of triaxial stress within the specimen. The ratio of the radial stress to the longitudinal stress at any given point on the notch cross-section is by definition the degree

Figure_14: da/dN vs $\sigma_{y.s.}$ for 4340 Steel, for $\Delta K = 20$ ksi $\sqrt{\text{in}}$; Pure Hydrogen Environment

of stress triaxiality at that point.

The location of the point of maximum stress triaxiality is dependent on the sharpness of the crack tip notch and the amount of plastic deformation in the area of the crack tip. The radial stress is greater for sharper notches and the amount of plastic deformation is minimal. The location of the point of maximum stress triaxiality can be roughly located by estimating the outer radius of the plastic zone along the notch cross-section. In the case of an extremely sharp crack tip the point of maximum stress triaxiality should lie only a very short distance directly ahead of the notch; while for a more blunt crack tip this point would lie substantially further into the interior of the specimen along the notch cross-section for the same applied stress.

Listed in Table II are the plastic zone size radii, r_y , computed for the three different strength levels at the various stress intensity levels at which the testing was done. These computations were obtained by using the following estimate for the plastic zone size radius:

$$r_y = \frac{K_I^2 (24)}{2\sigma_{y.s.}^2}$$

This then gives a rough estimate of the distance that the diffusing hydrogen must travel to reach the point of maximum stress triaxiality.

Also listed in Table II are the plastic zone size to specimen thickness ratios, r_y/t . This ratio is commonly used to designate the stress state existing in the specimen.

TABLE II

Plastic Zone Size Radii, Degree of Plane Strain and
Hydrogen Diffusion Times Computed for 4340 Steel

$\sigma_{y.s.}$ (psi)	ΔK psi $\sqrt{in}$	r_y (cm)	$\frac{r_y}{t}$	t_1 (sec)	t_2 (sec)	t_2 (min)
215,000	15,000	.0062	.01	.3	96	1.5
"	20,000	.0109	.03	1	297	5
"	25,000	.0173	.04	2.5	748	12
"	30,000	.0248	.05	5	1538	26
180,000	15,000	.0088	.02	.6	194	3
"	20,000	.0157	.03	2	616	10
"	25,000	.0246	.05	5	1513	25
"	30,000	.0356	.07	11	3168	53
"	40,000	.0635	.13	34	10081	168
"	50,000	.1060	.22	94	28090	468
"	60,000	.1397	.29	163	48700	812
"	70,000	.1905	.40	302	90726	1512
"	80,000	.2515	.53	527	158130	2635
150,000	20,000	.0226	.05	4	1277	21
"	30,000	.0508	.11	22	6451	108
"	40,000	.0914	.19	70	20885	348
"	50,000	.1422	.30	168	50552	842
"	60,000	.2032	.43	344	103226	1720
"	70,000	.2794	.59	650	195161	3253
"	80,000	.3556	.75	1054	316128	5269

According to current criteria⁽⁴⁹⁾ an $r_y/t > 1$ puts the material in a plane stress state while an $r_y/t < .2$ results in a plane strain situation.

Lastly the times required for the hydrogen to reach the point of maximum stress triaxiality are also computed and listed in Table II. They were arrived at by use of the well known diffusion equation:

$$x^2 = 4Dt$$

where: x = distance over which the atomic hydrogen must diffuse, cm

D = diffusivity of hydrogen, cm^2/sec , and

t = diffusion time

Two times were computed for each test. The first time, t_1 , was calculated by taking $D = 3 \times 10^{-5} \text{ cm}^2/\text{sec}$, the diffusivity of hydrogen in pure iron. The second time, t_2 , was obtained by taking $D = 10^{-7} \text{ cm}^2/\text{sec}$, a rough estimate of the diffusivity of hydrogen in an alloy steel, taking into account the possible encounters the hydrogen may have with voids and other trapping sites in the lattice.

It is hoped that the second time, t_2 , is a fairly reasonable estimate for the diffusivity of hydrogen in an alloy steel. However, the effect of the cycling stressing on the material is not taken into account in either t_1 or t_2 . Nor, is the fact that the crack tip is a moving rather than a stationary source of hydrogen.

The distance over which the hydrogen diffuses away from the crack tip can also be estimated from a slightly different

approach. Consider the crack tip to be a moving line from which hydrogen is diffusing. Using a diffusion equation, analogous to the heat conduction equation⁽⁵³⁾, in a moving frame of reference one obtains a hydrogen concentration profile proportional to

$$e^{-vx/2D}$$

where: v = velocity of moving frame of reference

x = distance ahead of crack tip

D = diffusivity of hydrogen

v is taken to be the average crack growth rate in fatigue, da/dt . A measure of the penetration distance of the hydrogen ahead of the crack tip is:

$$vx/2D = 1$$

The x values obtained for two diffusivity values, x_1 , for $D = 3 \times 10^{-5} \text{ cm}^2/\text{sec}$, the diffusivity of hydrogen in pure iron, and x_2 , for $D = 10^{-7} \text{ cm}^2/\text{sec}$, the diffusivity of hydrogen in an alloy steel, hopefully accounting for lattice trapping sites, are listed in Table III. These values of x represent the estimated distance that the hydrogen can diffuse in advance of the moving crack tip. The data was taken from the 180,000 psi material, at ΔK of 20 ksi $\sqrt{\text{in}}$.

The estimates which were made of the distance that the hydrogen can diffuse in advance of the moving crack tip obtained by the use of the moving frame of reference also lend support to the maximum stress triaxiality theory. The distances calculated from the estimated diffusivity of hydrogen in an alloy steel for the 180,000 psi yield strength

TABLE III

Hydrogen Diffusion Distances Computed for
4340 Steel at a ΔK of 20 $\text{ksi}\sqrt{\text{in}}$.

ΔK ($\text{ksi}\sqrt{\text{in}}$)	$\frac{da}{dN}$ (in/cycle)	Frequency (cps)	$\frac{da}{dt}$ (cm/sec)	x_1 (cm)	x_2 (cm)
20	5×10^{-6}	5	62.5×10^{-6}	.96	.0032
20	17×10^{-6}	.5	21×10^{-6}	2.8	.0095
20	38×10^{-6}	.05	4.8×10^{-6}	12.5	.042

material at a $\Delta K = 20 \text{ ksi } \sqrt{\text{in}}$ (see Table III) are of the same order of magnitude as the r_y values computed for the 180,000 psi material at the lower ΔK levels; (see Table II).

Consider in this light the results of the present investigation beginning with the data obtained with the 215,000 psi yield strength material. The crack tip at this yield strength is relatively sharp and the plane strain situation minimizes the plastic deformation in the crack tip region so that the point of maximum stress triaxiality should lie only a very short distance in front of the notch. It would be expected that the severity of the hydrogen induced failure would be a maximum at this yield strength because the hydrogen would have to diffuse only a very short distance. This was indeed the case.

The severity of the hydrogen damage increased as the test frequency decreased. Babcock also observed this behavior; he proposed that at the higher frequencies there is less time per cycle available for the hydrogen to diffuse to the point of maximum stress triaxiality. This caused the reduction in the environmental crack growth rate. It is suggested that the same explanation is applicable in the present case.

The same line of reasoning can be extended to the data obtained with the 150,000 and 180,000 psi yield strength specimens. The sharpness of the crack tip decreases as the yield strength of the material decreases due to the increase

in the amount of plastic deformation caused by the lack of a plane strain situation. Thus, the lower the yield strength of the material, the further away from the crack tip is the point of maximum stress triaxiality. See Table II. This means that the hydrogen must diffuse further in the low than in the high yield strength material to reach the critical region of maximum stress triaxiality. In addition to this, the lower yield strength material has an inherent toughness of its own due to its ability to plastically deform. Hence, the observed decrease in the severity of the hydrogen induced cracking in the 150,000 and 180,000 psi yield strength 4340 steel.

The gradual diminishing of the frequency dependence and finally its total disappearance in the lowest yield strength material may possibly be due to the increased distance that the hydrogen must diffuse. Although the diffusion times listed in Table II are only rough estimates and although they do not reflect the effect of the cyclic stressing of the material, the difference in the times for diffusion in the two extreme yield strengths are substantial.

Another way of categorizing the severity of the hydrogen induced cracking is by noting the exponent n in the proposed fatigue crack propagation expression, Equation (4). The larger n is, the more severe is the damaging effect of the hydrogen. The n values obtained in the present experiment for the plots in Figures 11, 12 and 13 are listed in Table IV. The crack propagation rates for the lowest yield

TABLE IV

Tabulated n Values from Figures 11, 12 and 13

$\sigma_{y.s.}$ (psi)	Environment	Freq. (cps)	n
150,000	H ₂	.05	3.45
"	H ₂	.5	3.45
"	H ₂	5	3.45
"	.4% by vol. O ₂ -H ₂	.05,.5,5	3.0
"	Ar	.05,.5	3.0
180,000	H ₂	.05	6.7
"	H ₂	.5	7.1
"	H ₂	5	6.7
"	.4% by vol. O ₂ -H ₂	.05,.5,5	3.4
"	AIR	.05,.5,5	3.4
215,000	H ₂	.05	19.1
"	H ₂	.5	14.3
"	H ₂	5	11.4
"	.4% by vol. O ₂ -H ₂	5	19.1
"	.4% by vol. O ₂ -H ₂	.05,.5	-
"	Ar	.05,.5	3.5

strength specimens at the three test frequencies all lie on the same straight line with an n equal to 3.40. In the intermediate yield strength specimens the crack growth rates at .05, .5 and 5 cycles per second fall on straight lines with n 's equal to 6.7, 7.1 and 6.7 respectively. The frequency dependence of the crack growth rate at this particular yield strength cannot be discerned from the n values. Rather it is observed as an upward displacement of the crack growth curves at the lower frequencies. The crack growth rate curves for the 215,000 yield strength specimens at .05, .5 and 5 cycles per second produced n values of 19.1, 14.3 and 11.4. Obviously the hydrogen was most damaging in these high yield strength specimens. The increased severity of the hydrogen accelerated crack growth due to the decrease in frequency is reflected in the increased values of n at the two lower frequencies. Only in the case of the 215,000 psi yield strength material was there a significant variation of n with frequency.

B. Fatigue Crack Growth in a .4% by Volume
Oxygen-Hydrogen Mixture

The very pronounced protective effect of the .4% by volume oxygen-hydrogen mixture is shown in Figures 11, 12 and 13. This particular concentration of oxygen was selected because it was the lowest amount of oxygen that had been reported to stop hydrogen induced cracking under sustained loading conditions⁽³⁸⁾.

The beneficial effect of this very small amount of oxy-

gen can be clearly seen. The fatigue crack growth rates obtained from the two lower yield strength materials at .05, .5 and 5 cycles per second fall on nearly identical straight lines which lie well to the right of the lines obtained in pure hydrogen. No frequency dependence of the crack growth rates could be discerned. The 150,000 psi yield strength 4340 had a crack propagation law exponent, n , equal to 3.0 while the 180,000 psi yield strength 4340 had an n equal to 3.4. Fatigue crack growth rates for the 150,000 and 180,000 psi yield strength specimens were 2.7×10^{-6} and 3.0×10^{-6} in/cycle at $\Delta K = 30 \text{ ksi } \sqrt{\text{in}}$ and 34×10^{-6} and 50×10^{-6} in/cycle at $\Delta K = 70 \text{ ksi } \sqrt{\text{in}}$. Within the limits of experimental error these crack growth rates are nearly identical.

The testing of the 215,000 psi yield strength specimens in the .4% by volume oxygen-hydrogen mixture produced somewhat more complicated results. A crack growth rate plot could be made only from the tests run at a frequency of 5 cycles per second. This plot, shown in Figure 13, deviates from a straight line at the higher stress intensity levels where it also displays a tendency to become coincident with the 5 cycles per second curve obtained in pure hydrogen. Crack growth in the oxygen-hydrogen mixture was not observed until a stress intensity factor level of $16 \text{ ksi } \sqrt{\text{in}}$ was reached. Apparently the oxygen-hydrogen mixture was completely protective until a ΔK level of $16 \text{ ksi } \sqrt{\text{in}}$ was reached, after which it was only partially protective.

Crack growth rate curves could not be drawn for the

215,000 psi yield strength materials tested at .05 and .5 cycles per second because not enough data could be obtained. At both frequencies no crack growth would occur until a ΔK of 22 ksi $\sqrt{\text{in}}$ was reached. Subcritical crack growth then occurred between stress intensity levels of 22 and 26 ksi $\sqrt{\text{in}}$. At stress intensity factors greater than 26 ksi $\sqrt{\text{in}}$ the crack propagated in a catastrophic fashion making accurate crack growth measurements impossible. The crack growth rate measurements made between $\Delta K = 22$ and 26 ksi $\sqrt{\text{in}}$ were not very reproducible as can be seen from Figure 13.

The 180,000 psi yield strength material was also tested in room air at .05, .5 and 5 cycles per second and the resultant crack growth rate plot is shown in Figure 12. This plot shows the crack growth rate to be frequency independent and coincident with the curve obtained in the .4% by volume oxygen-hydrogen mixture. Evidentially the protective influence of the small oxygen addition is equivalent to that of an argon or room air environment in the case of the two lower yield strength materials at all the test frequencies.

These results illustrate the protective influence of the .4% by volume oxygen addition. The precise mechanism of this protective effect though is not unambiguously clear. It was suggested by Hancock and Johnson⁽⁴¹⁾ that oxygen was preferentially adsorbed at the crack tip with the subsequent formation of an oxide layer which would not allow the passage of the damaging hydrogen. This same line of reasoning

will be used in treating the present results.

The protective influence of the small oxygen addition can be explained in terms of the preferential adsorption of an oxide layer preventing the damaging hydrogen adsorption. In the case of the 215,000 psi yield strength 4340 for the .05 and .5 cycle per second tests the protection of the oxide layer extended up to a ΔK of approximately 22 ksi $\sqrt{\text{in}}$ while in the tests at 5 cycles per second sub-critical crack growth was observed above a ΔK of approximately 16 ksi $\sqrt{\text{in}}$. In the inert argon atmosphere no crack growth was observed below a ΔK of approximately 22 ksi $\sqrt{\text{in}}$.

It is proposed that a cyclic breakdown and reformation of the adsorbed oxide layer occurs somewhere in the stress intensity factor range of 16 to 22 ksi $\sqrt{\text{in}}$. In the lower frequency tests where more time per cycle is available for hydrogen diffusion, the critical concentration of hydrogen was able to reach the point of maximum stress triaxiality very rapidly during the time the oxide film was ruptured. There was only a small amount of subcritical crack growth once the film had ruptured because the stress intensity level was high enough to cause rapid fracture. However, in the testing at 5 cycles per second less time per cycle was available for the hydrogen to diffuse resulting in somewhat smaller concentrations of hydrogen at the critical point of triaxiality allowing subcritical crack growth to occur, but at a slower rate than observed in the pure hydrogen atmosphere at the same stress intensity level.

The protective influence of the .4% by volume oxygen addition in the case of the lower yield strength materials can be explained in a similar fashion. It is proposed that the adsorbed oxide layer ruptures and reforms cyclically somewhere between $\Delta K = 16$ and 22 ksi in as observed in the 215,000 psi yield strength material. However, the hydrogen must diffuse over a significantly longer distance to reach the point of maximum stress triaxiality. This distance would increase as the stress intensity factor range was increased because the plastic zone size will increase. The impediment provided by the oxide layer is sufficient to prevent the critical concentration of hydrogen from reaching the region of maximum stress triaxiality rapidly enough to be the rate controlling factor in the crack growth process.

Thus, before the critical concentration of hydrogen can reach the point of maximum stress triaxiality, the stress conditions ahead of the crack tip would become sufficiently severe to cause the crack to propagate as it would in a non-aggressive environment. A new crack tip is then formed with the subsequent movement of the point of maximum stress triaxiality to a new location ahead of the crack tip. This process would then repeat itself as the stress intensity factor range was increased throughout the testing. This proposal is supported by the coincidence of the .4% by volume oxygen-hydrogen curves, the argon curves and the room air curve for the lower yield strength materials.

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The establishment of a stress intensity value between 16 and 22 ksi $\sqrt{\text{in}}$ at which the oxide layer ruptures is not unreasonable. Both Hancock and Johnson⁽⁴¹⁾ and Sawicki⁽³⁸⁾ reported that small oxygen additions to a pure hydrogen atmosphere prevented environmentally induced crack growth in H-11 steel under sustained loading. Hancock and Johnson⁽⁴¹⁾ observed this phenomena at a K-level of 18.5 ksi $\sqrt{\text{in}}$ and Sawicki⁽³⁸⁾ over a K-level range of 15 to 25 ksi $\sqrt{\text{in}}$. Unfortunately no attempt was made to determine at what stress intensity factor level the beneficial effect of the oxygen was no longer present.

C. Crack Propagation Laws

As summarized in Part I numerous attempts have been made to describe fatigue crack growth rate data by a power law relationship of the form shown in Equation (4). With the exception of the 5 cycles per second tests in pure hydrogen and in the oxygen-hydrogen mixture with the 215,000 psi yield strength material whose $\ln da/dN$ vs $\ln \Delta K$ plots showed a deviation from linearity at the higher stress intensity levels, all of the present data indicated a power law relationship similar to that described by Equation (4) at all the K-levels at which the testing was done.

The exponents in the power law, n , obtained from the slopes of the $\ln da/dN$ vs $\ln \Delta K$ plots for the various test conditions are listed in Table IV. From this table it can be clearly seen that for each different yield strength and

test condition a different value of n was obtained.

Consider the n 's obtained over the range of yield strengths in the non-aggressive environments; i.e., the .4% by volume oxygen-hydrogen mixture (with the exception of the 215,000 psi yield strength data), the room air and the inert argon. The non-aggressive environment curves lie on coincident straight lines for each yield strength. The n values ranged from 3.0 for the 150,000 psi yield strength 4340 to 3.4 for the intermediate yield strength 4340 to 3.5 for the 215,000 psi yield strength material. This slight increase in n reflects the decrease in the fracture toughness of the material as the stress state changes from plane stress to plane strain.

This data does not suggest an invariant power law for 4340 steel. Hence, it is suggested that any attempts to describe the crack growth behavior of a wide variety of materials with an invariant power law should be subjected to a great deal of scrutiny.

However, the results of Miller⁽⁴⁷⁾ prove to be of interest; he obtained n values in room air with 4340 steel which had been tempered at various temperatures. The exponent n was reported to be inversely related to the fracture toughness of the material. A similar trend was found in this investigation. Miller⁽⁴⁷⁾ obtained n values of 3.29 and 3.17 for 4340 steel heat treated to yield strengths of 220,000 and 160,000 psi. In this present investigation n values of 3.5 and 3.0 were obtained for 4340 tempered to

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yield strengths of 215,000 and 150,000 psi. Taking into account the experimental error involved in both these investigations this apparent correlation is quite good.

Now consider the n values obtained in the pure hydrogen environment. Obviously as the yield strength of the material increases so does the severity of the hydrogen induced cracking. This is reflected in an increase in the n values. Certainly n is not an invariant integer. As noted previously, only in the testing of the highest yield strength 4340 was the frequency dependence of the crack growth rate reflected by the values of n .

The present data could not be analyzed in the manner suggested by Wei and Landes⁽⁴⁹⁾. This was due to the unavailability of sustained testing crack growth rate data for 4340 steel obtained in a pure hydrogen environment. This was unfortunate because the superposition approach of Wei and Landes is the only attempt to systematically predict fatigue crack growth rates in different aggressive environments under different loading conditions. If proven correct it would be an invaluable asset in predicting structural load bearing capabilities for high strength steels under a variety of loading and environmental conditions.

CHAPTER IV. CONCLUSIONS

(1) The deliterious effect of a hydrogen environment on the fatigue crack growth behavior in a high strength steel is evident. The severity of the observed hydrogen induced cracking decreases as the yield strength of the material decreases. The fatigue crack growth behavior in hydrogen is frequency dependent in the 215,000 and 180,000 psi yield strength 4340 steel. The higher the frequency, the slower the fatigue crack growth rate. There is no frequency dependence of the fatigue crack growth rates in hydrogen in the 150,000 psi yield strength 4340.

(2) A small oxygen addition to a pure hydrogen environment has a beneficial effect on fatigue crack growth behavior in 4340 steel. A .4% by volume oxygen addition to a pure hydrogen atmosphere prevents hydrogen induced cracking of 215,000 psi yield strength 4340 up to a ΔK level of 16 ksi $\sqrt{\text{in}}$ at a test frequency of 5 cycles per second. Above this ΔK level the fatigue crack growth was retarded. No fatigue crack growth occurs in the .4% by volume oxygen-hydrogen mixture up to a ΔK level of 22 ksi $\sqrt{\text{in}}$ in the 215,000 psi yield strength 4340 at frequencies of .05 and .5 cycles per second. Above this ΔK level crack growth is rapid and fast fracture occurs. In pure hydrogen, crack growth occurs above a ΔK of 12 ksi $\sqrt{\text{in}}$ with fast fracture usually occurring between a ΔK of 20 to 25 ksi $\sqrt{\text{in}}$.

Hydrogen induced fatigue cracking is retarded in 150,000 and 180,000 psi yield strength 4340 steel at all the test frequencies by the small oxygen addition. The preventive effect of the oxygen addition is equivalent to that of pure argon and room air atmospheres at these yield strengths. No frequency dependence of the crack rates is detectable in the two lower yield strength materials in the oxygen-hydrogen mixture.

(3) The usefulness of a fatigue crack propagation law with a invariant integer exponent is limited. The exponent, n , varies with the test frequency, the yield strength of the material and the test environment in 4340 steel. It can be used, however, in comparing the severity of fatigue crack growth behavior in the same material by varying one test condition and holding the others constant.

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