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THE EFFECT OF HYDROGEN ON METALS

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ABSTRACT - Hydrogen embrittlement of metals is separated into three categories: embrittlement resulting from blister formation, embrittlement from hydride formation, and generalized hydrogen embrittlement. The causes and mechanisms are discussed, and guides to select materials to resist this phenomenon are included.

INTRODUCTION

Hydrogen embrittlement and/or hydriding has led to failures of fuel cladding in nuclear reactors,¹ breakage of aircraft components,² leakage from gas-filled pressure vessels used by NASA,³ delayed failure in numerous high-strength steels,⁴ reductions in mechanical properties of nuclear materials,⁵ and blisters or "fish-eyes" in copper,⁶ aluminum,⁷ and steel parts.⁸ The role of hydrogen in stress corrosion cracking of titanium,⁹ aluminum,¹⁰ and both austenitic¹¹ and high-strength¹² steels has also been emphasized. The importance of these considerations is readily apparent in the current literature; for example, hydrogen was mentioned in the title of ~5% of the recent articles in *Metallurgical Transactions* and in ~10% of those in *Corrosion*. Furthermore, there have been at least three international conferences on hydrogen/metal interactions within the past two years.

The effects of hydrogen on metals depend on several factors including:

- Hydrogen solubility and diffusivity
- The possibility of reaction to form a hydride
- The possibility of reaction between hydrogen and impurity and/or alloying elements.

Although there are several proposed mechanisms by which hydrogen may affect the load-bearing capabilities of metals, most of them require the development of a critical hydrogen concentration within the metal lattice. This critical concentration is proposed to lower lattice cohesion,¹³ cause localized plasticity,¹⁴ restrict glide,¹⁵ precipitate as gas bubbles,¹⁶ or cause embrittlement by a combination of these and other mechanisms.¹⁷⁻²⁷ Regardless of the mechanism, however, the fact that hydrogen degrades mechanical properties is well-known, and hydrogen embrittlement has been extensively studied. These studies have produced a degree of understanding which often permits design engineers to choose materials so as to minimize the probability of adverse hydrogen effects. The objective of this review is to emphasize this understanding.

HYDROGEN EFFECTS

Blistering

Hydrogen can be absorbed by a metal during casting, forging, heat treatment, welding, and finishing, or in service. If the external hydrogen pressure is reduced after absorption, hydrogen diffusion from the metal is required for the system to obtain thermodynamic equilibrium. The internal hydrogen concentration, C , in equilibrium with an external hydrogen pressure, p , over a hydride-free metal is

$$C = C_0 p^{1/2} \exp(-\Delta H_s/RT) \text{ cc gas (NTP)/cc metal}$$

Hydrogen diffusion from the metal is also required for equilibrium if the temperature is reduced. If the temperature change is rapid, diffusion may be too slow to maintain equilibrium and the metal lattice may become supersaturated with hydrogen.

Local equilibrium may be maintained within the lattice, however, by precipitation of molecular hydrogen at internal cavities (or defects). The hydrogen pressure at these cavities will depend on the initial concentration and the amount of the temperature decrease; it can easily exceed the pressure required to strain the metal plastically. Such strain opens microcavities and causes blistering. For practical purposes, however, blisters typically do not originate because of exposure to hydrogen gas but are developed by reactions between the metal and water vapor during melting and heat treatment, or from hydrogen introduced during finishing operations.

Iron and aluminum will react with atmospheric moisture at elevated temperatures according to the equations



The metal lattice becomes supersaturated with hydrogen because the effective atomic hydrogen pressure during the reaction is many times the equilibrium pressure. Subsequent precipitation of molecular hydrogen leads to the development of high pressures and blisters (Figure 1a). If hydrogen is absorbed while the metal is liquid, gas bubbles (blow holes) may develop because of the decrease in hydrogen solubility during solidification (Figure 1b). Precipitation of hydrogen gas at internal defects has also led to blistering after acid pickling, electroplating, and other finishing operations.

The cause of blistering is well-known: a change in environment (temperature, hydrogen over-pressure) leads to precipitation of hydrogen gas from a supersaturated metal lattice. Because this mechanism is well-known, handling and finishing techniques have been developed to minimize this form of embrittlement. Vacuum melting and degassing minimize the quantity of hydrogen in the metal. Acid pickling and other such processes which may introduce hydrogen are avoided when practical, and possible moisture sources, such as the coatings of welding electrodes, are carefully considered before use. Melts are degassed by bubbling an active or inert gas through the liquid (chlorination of aluminum is an effective method of removing hydrogen, and hydrogen is removed from some metals, such as magnesium, simply by holding the metal molten to allow hydrogen outgassing before casting.

Inclusions and other defects are typical sites for hydrogen precipitation. These sites can be controlled, to a limited extent, by processing techniques. For any given hydrogen content, the extent of blistering decreases when the number of precipitation sites is increased. Thus a homogenous distribution of very small inclusions will minimize susceptibility to embrittlement. In addition to this indirect effect of inclusions on embrittlement, hydrogen may diffuse into a metal and react directly with inclusions or alloy phases²⁸⁻³⁰ to form a gaseous product. This phenomenon is often termed "hydrogen sickness."

Hydrogen sickness in copper is more correctly described by the term "steam embrittlement" and is only observed in copper which contains oxygen. It is caused by hydrogen reduction of copper oxide inclusions and the formation of water. The problem was encountered originally when hydrogen gas was used to maintain a reducing atmosphere during the heat treatment of copper. Hydrogen diffused into the metal during such anneals and reacted with oxide particles to form water. At temperatures above the critical temperature for water, the steam pressure generated by the reaction often exceeded the strength of the copper and caused plastic deformation and/or tearing. Oxide inclusions in copper alloy are often found along grain boundaries; therefore, examination of steam-embrittled copper frequently reveals grain boundary cracks or cavities (Figure 2). However, cavities are also found within the grains and along prior grain boundaries in alloys that have received thermo-mechanical treatments after oxide precipitation. Problems relating to the occurrence of steam embrittlement can be avoided if:

- Alloys chosen for hydrogen service are free of oxide inclusions.
- Alloys containing oxide are not annealed in hydrogen.

Silver is also susceptible to steam embrittlement; however, in this case oxide reduction, *per se*, may not occur. Blisters have been found on high purity (99.999⁺ percent) silver annealed in hydrogen.²⁹ Hydrogen diffuses into the silver and reacts with the dissolved oxygen to form water vapor, which is nearly insoluble and therefore precipitates. Detailed metallographic studies indicate that grain boundary precipitation is predominant. Individual bubbles continue to grow as the reaction proceeds until bubble agglomeration leads to grain boundary cracking.

Blistering is also often observed in steels used in the chemical and petrochemical industry because of hydrogen reaction with carbide inclusions and carbon in solid solution.¹⁹ The reaction is termed "hydrogen attack," and high pressures are developed because of methane formation. The formation of methane reduces the mechanical properties because in addition to the formation of high pressure gas bubbles, the carbides and dissolved carbon are eliminated, thus lowering the strength. The susceptibility to hydrogen attack can be reduced by the use of steels containing titanium, vanadium, or other additions to react with carbon to form carbides that are not reduced by hydrogen.

Hydrogen sickness, hydrogen attack, blistering, and other forms of damage caused by hydrogen reaction with alloy or impurity elements are generally described as irreversible hydrogen embrittlement. Although some ductility may be restored by heat treatments, the removal of strengthening phases, such as carbides, cause strength losses which are not generally recovered. Furthermore, it is doubtful that the microvoids, pores, blisters, or fissures developed during the attack can be healed effectively by thermomechanical treatments. Therefore, for design or fabrication considerations, prevention is a much more effective measure than subsequent attempts to recover strength and ductility losses.

Hydride Embrittlement

Zirconium, titanium, niobium, vanadium, uranium, and other such metals and their alloys form hydrides if the external hydrogen pressure is high enough. The titanium-hydrogen phase diagram, Figure 3, is typical of titanium, zirconium, and hafnium, and shows that below the eutectoid temperature, hydrogen solubility decreases rapidly with decreasing temperature and is practically nil at room temperature. The other hydride formers show similar solubility decreases. Because of this decrease in hydrogen solubility, hydrides often precipitate even when the hydrogen content of the metal is very low. In some respects this is similar to blistering; a hydrogen-rich phase is formed, and this phase may affect the mechanical properties of the metal.¹⁷ Under severe hydrogen charging conditions, the strains developed by precipitation of the less-dense hydride phase can produce stresses sufficient to cause failure even when no load is applied.¹⁹

Susceptibility to hydride embrittlement is dependent on numerous variables including: hydrogen content, hydride distribution and morphology, temperature, and strain rate. The metal hydrides are not only brittle and less dense than the metal matrix, but precipitate at dislocations,³¹ at grain and twin boundaries,³² and at other defects. This combination of hydride-matrix properties influences both the initiation and growth stages of fracture.³²

Hydride embrittlement generally increases with increasing hydrogen content (Figure 4). This increased susceptibility is due to an increase in the number of hydrides per unit volume, a corresponding decrease in the interhydride spacing, and a change in actual hydride distribution.³³ Hydride distribution (orientation), which also affects interhydride spacing, is of such major importance in some cases that the role of hydrogen content is suppressed. Tensile specimens of Zircaloy-2 containing hydrides oriented parallel to the stress axis failed after 10% elongation; whereas comparable specimens containing hydrides perpendicular to the stress axis, but with only 0.05 times the total hydrogen, exhibited no macroscopic ductility.³⁴

Increasing the temperature decreases the susceptibility to hydride embrittlement because, for any given hydrogen content, the volume fraction of hydride decreases. However, there is some evidence that above a critical temperature, the hydride may become ductile,³⁵ and thus be less likely to enhance crack initiation. Ductile-to-brittle transitions have been observed in hydrided zirconium alloys,³³ and the transition temperature has been shown to increase with increasing hydrogen content (Figure 5). This phenomenon was shown to be consistent with variations in interhydride spacing and led to the conclusion that the embrittling effect of hydride precipitation is due to cracking along the hydride.³³ The shape and distribution of the hydrides are strongly dependent on the heat treatment prior to precipitation and on the cooling rate during precipitation.³⁶ Because cracks propagate along metal-hydride interfaces (Figure 6), cooling rates also influence embrittlement. Slow cooling apparently promotes the formation of thin platelets primarily at grain boundaries and on specific crystallographic planes, whereas rapid quenching precipitates highly dispersed particles. Correspondingly, the toughness (resistance to hydride embrittlement) is higher in hydrided titanium when the hydrides are compact, than when thin hydride platelets are present.³⁷

The existence of hydride precipitates before testing is not a necessary condition for hydride embrittlement. Vanadium alloys with hydrogen contents less than the apparent terminal solid solubility are embrittled because the test stresses (strains) cause hydride precipitation and subsequent embrittlement.³⁸ In such case, the apparent solubility may be decreased by an applied stress because precipitation of the low-density hydride phase is accompanied by nucleation of numerous dislocations at hydride-matrix interfaces.³⁶ Test stresses aid hydride nucleation and thus lower the apparent solubility. This effect is markedly different from the thermodynamically predicted increase in actual hydrogen solubility through the application of elastic tensile stresses.³⁹

Surface hydrides also affect the mechanical properties of hydride-forming metals. Failure of tensile specimens with surface hydride layers is initiated in the hydride and occurs at very low strains. Most of the conditions for surface hydride formation are met by exposure of hydride-forming metals to gaseous hydrogen, even at very low pressures. For example, calculations show that for alpha-phase titanium at room temperature, the equilibrium pressure above which hydrides will form is 5×10^{-14} torr. However, titanium alloys can be exposed to much higher hydrogen pressure with no adverse effects because of the protection afforded by the oxide film typically present on such alloys.⁴⁰ Surface hydride formation can be predicted from a knowledge of hydrogen transport rates in both the oxide film and the metal matrix.⁴⁰ Furthermore, surface hydride formation, *per se*, was shown to degrade the mechanical properties and may (without the application of stress) cause disintegration of the specimen (Figure 7).

Hydride embrittlement, in most cases, is a maximum at high strain rates. Studies with hydrided titanium alloys⁴¹ show that impact and notch specimen testing are effective methods for determining the susceptibility to hydride embrittlement (Figure 8).

Generalized Hydrogen Embrittlement

Failure Under Static Conditions

Delayed failure (static fatigue) is a time-dependent form of hydrogen embrittlement that occurs primarily in high strength steels.² Hydrogen-charged specimens, loaded to stresses less than the normal fracture stress, often fail after an incubation time depending upon temperature, hydrogen content, and other test variables. This phenomenon has been extensively studied primarily with notched specimens,⁴² and is perhaps best illustrated by Figure 9. The upper critical stress is the normal fracture stress of the alloy, and the lower critical stress is the stress below which failure will not occur. The incubation period is the time required for the initial crack formation or the onset of irreversible damage, while the fracture time is simply the time to complete rupture. The lower critical stress and the failure time increase rapidly with increasing notch root radius⁴³ and are sensitive functions of temperature, hydrogen content, and specimen strength. The incubation time is reversible, and measurements of the kinetics of reversibility shown an activation energy of approximately 9000 cal per mole.⁴⁴ This activation energy has been used to support and/or discredit several proposed mechanisms for hydrogen embrittlement. However, because the kinetics of hydrogen transport in steels are severely complicated by trapping,⁴⁵ use of this type of kinetic data to support any mechanism is questionable.^{14,22}

Delayed failure is one of the most insidious forms of hydrogen embrittlement because it can occur without warning in parts which are not being exposed to hydrogen. The hydrogen absorbed during a pickling or finishing operation can cause delayed failure during subsequent use even when nominal test procedures fail to indicate hydrogen damage. For this reason, tests to determine susceptibility to delayed failure were developed in the 1950's (ASTM Standard E-8-65T). At the June 1972 ASTM Committee F-7.04 meeting on "Standardization of Testing in Hydrogen" numerous testing techniques were reported for establishing delayed failure susceptibility; however, the results of the different techniques were not necessarily comparable. The difference in results show that qualification of a particular plating process, finishing operation, or pickling bath may be as dependent on the test technique as on the process itself. For this reason, it is desirable that the susceptibility of a part to delayed failure or the qualification of a particular finishing operation be determined through test procedures recommended in the ASTM Standards rather than through tests described in ASTM STP's but not yet accepted as standard techniques.

Failure Under Dynamic Conditions

One of the early examples of hydrogen embrittlement⁴⁶ shows the temperature and strain-rate dependence of ductility losses in cathodically-charged mild steel (Figure 10). These and similar data show that the degree of embrittlement

- increases with increasing hydrogen content,⁴⁷
- is a maximum at an intermediate temperature,⁴⁶ and
- decreases with increasing strain rates.⁴⁶

Embrittlement in hydrogen-charged austenitic steels,¹⁶ nickel,⁴⁸ and high-strength steels² shows similar characteristics. Furthermore, studies of the effects of temperature, strain rate, and hydrogen content on hydrogen transport by dislocations⁴⁹ suggest such transport is of major importance in hydrogen embrittlement processes. This conclusion led to the following phenomenological description⁵⁰ of hydrogen embrittlement in nonhydride-forming metals:

"The deleterious effects of hydrogen on the tensile properties of metals are caused by the association and movement of hydrogen with dislocations. Hydrogen-dislocation interactions modify plastic deformation processes by stabilizing microcracks, by changing the work-hardening rate, and by solid-solution hardening."

This description coupled with the concept of "hydrogen-binding-to-dislocations" provides a basis for explaining the temperature, strain rate, and hydrogen content dependence of hydrogen embrittlement. The amount of hydrogen transported, C_T , is related to dislocation movement into any region of a specimen during a dynamic test through the equations

$$C_T \propto \rho_m C_L \bar{V}$$

and
$$C_L = C \exp(-G_B/RT)$$

where C is the nominal hydrogen content; C_1 , the hydrogen content at a dislocation core; ρ_m , the mobile dislocation density; \bar{V} , the average dislocation velocity; and G_B , the hydrogen-dislocation binding energy, and is negative when binding occurs.

Susceptibility to hydrogen embrittlement is a maximum at some intermediate temperature and apparently vanishes at elevated temperatures, even under conditions of high temperature creep and stress rupture,⁵¹ but embrittlement has been observed in tests as low as 77°K.⁴⁴

The decreasing susceptibility with increasing temperature at any given hydrogen concentration is caused by an exponential decrease of the ratio C_1/C with temperature. The ratio approaches 1 as T increases, and, for a given amount of dislocation motion, less hydrogen is transported at the higher temperature. On the other hand, the amount of hydrogen transported increases with decreasing temperature until the dislocations become saturated (i.e., $C_1 = 1$), and/or until the lower temperature reduces hydrogen mobility to the point where dislocations must break away from their hydrogen atmospheres in order to move.

Most proposed embrittlement mechanisms require that (a) a critical, or threshold, amount of hydrogen be localized, and (b) the amount of embrittlement is proportional to the amount of hydrogen localized. Therefore, for a given strain rate and hydrogen content, C_T decreases as T increases at any temperature above the dislocation saturation temperature. At temperatures below this temperature, C_T is limited by the diffusivity of hydrogen and decreases as T decreases. Thus, a maximum in the amount of hydrogen transported and localized is obtained at an intermediate temperature, and, hence, a maximum in embrittlement is observed. The temperature of this maximum should increase with increasing strain rate and increasing hydrogen content as has been observed in experimental studies.⁴⁶

The amount of hydrogen transported per unit strain decreases as the strain rate increases (Figure 11 and Reference 49). This dependency, coupled with the dislocation-transport model for embrittlement, predicts the observed decrease in embrittlement susceptibility with increased strain rate.

For any given amount of dislocation motion, the amount of hydrogen transported is directly proportional to the hydrogen content until the dislocations become saturated with hydrogen. This proportionality indicates that the strain-to-fracture should decrease linearly with hydrogen content until the dislocations are saturated. Further increases in hydrogen content above this saturation value should have little effect on hydrogen embrittlement. This behavior has been experimentally observed (Figure 12 and Reference 47).

In addition to the effects of hydrogen content, temperature, and strain rate on the susceptibility to hydrogen embrittlement, studies have also shown that embrittlement is promoted by coplanar dislocation motion⁵ and that susceptibility to embrittlement is greater for high-strength alloys than for low-strength alloys of similar structures (Figure 13 and Reference 25). These embrittlement studies provide some simple criteria for alloy selection to minimize embrittlement susceptibility. Because a critical quantity of hydrogen is apparently required for embrittlement, a low hydrogen solubility is desirable. For example, neither copper or aluminum are susceptible to dynamic embrittlement, and their hydrogen solubilities are much lower than readily embrittled metals such as alloys of iron and nickel. Embrittlement is promoted by low stacking-fault energies, high negative dislocation-hydrogen binding energies, and high strengths; thus, alloys having these general properties should be avoided.

Alloy design can also minimize embrittlement. For example, fine dispersions of incoherent precipitates have been shown to be effective in eliminating some adverse hydrogen effects in nickel,⁵² and "prestraining" in air before hydrogen exposures have minimized embrittlement in other alloys.⁵³

Failure in Hydrogen Environments

Hydrogen environment embrittlement is embrittlement which results when an initially "hydrogen free" metal is deformed in a hydrogen environment. This phenomenon was first described in the mid-1950's^{53,54} in studies which showed that ductility of many metals tested in high pressure gaseous hydrogen was significantly less than expected. Subsequent studies⁵⁵ indicated that "hydrogen is supplied to the root of the propagating crack both from the metal and directly from the gas phase." Other investigators^{3,4} confirmed these results and suggested that hydrogen environment embrittlement is distinctly different from other embrittlement modes. This distinction probably developed because tests in high pressure hydrogen were often found to be independent of exposure time.²¹ This independence was interpreted as showing that diffusion and hence absorption are not vital to hydrogen environment effects. However, it has recently been shown^{49,50} that plastic deformation during hydrogen exposure greatly affects absorption, and that both the depth of significant absorption and the amount of hydrogen absorbed are significantly greater than predicted by diffusion theory. Thus, absorption readily occurs during environmental embrittlement, and there is little need to invoke a different embrittlement mode for this phenomenon. Furthermore, the effects of temperature, strain rate, and hydrogen content (pressure) on susceptibility to hydrogen embrittlement have been shown to be the same in both internal (precharged) and environmental tests.¹⁶ These results suggest that a single mechanism accounts for both forms of embrittlement. Nevertheless, environmental hydrogen embrittlement is now a well-studied phenomenon, and the compatibility of numerous metals with high pressure hydrogen has been established.³ These results indicate that selection of materials for gaseous hydrogen exposures should be based on criteria similar to those for minimizing the effects of internal hydrogen.

Hydrogen environments also affect slow crack growth rates for precracked specimens⁵⁶ and the fatigue life² of many metals. Crack growth in H-11 steel has been shown to be a sensitive function of environment. The stress intensity for unstable crack growth was ~ 11 kpsi $\sqrt{\text{in.}}$ in hydrogen and was ~ 40 kpsi $\sqrt{\text{in.}}$ in dry argon. Furthermore, hydrogen-induced, slow-crack growth in compact tension specimens could be stopped if 0.6% oxygen was added to the atmospheric hydrogen environment (Figure 14) showing that oxygen is a remarkable protective agent against hydrogen-induced cracking. Other studies³ have shown that oxygen impurities will inhibit cracking of steels in high pressure hydrogen. The mechanism for this inhibition (or protection) is not established, but is probably related to oxygen inhibition of hydrogen absorption.⁵⁰

PROPOSED MECHANISM FOR GENERALIZED EMBRITTLEMENT

The deleterious effects of hydrogen on the mechanical properties of nonhydride-forming metals are caused by the association and movement of hydrogen with dislocations. This movement transports hydrogen to high strain regions in the metal lattice and leads to localized, high-hydrogen concentrations. Breakdown of dislocation pileups and dislocation egression into microvoids, or other processes which lead to dislocation annihilation cause local supersaturation of the lattice. Under these conditions, hydrogen can precipitate to nucleate and/or stabilize microvoids. Furthermore, the association of hydrogen with dislocations affects, dislocation dynamics, and thus the association, *per se*, contributes to embrittlement.

The dislocation-transport model requires that plastic deformation accompany embrittlement. This model agrees with the hydrogen-assisted cracking model.¹⁴ Another basic premise of the proposed model is that localized lattice supersaturation must precede embrittlement. This premise agrees with the generalized model for hydrogen embrittlement, which assumes that embrittlement results from precipitation of a hydrogen-rich phase whose mechanical properties differ from those of the matrix.¹⁷ The high localized concentrations are promoted by coplanar dislocation motion, large negative hydrogen-dislocation binding energies, and high hydrogen concentrations, either in the specimens or in the test environment. The mechanism by which the regions of lattice supersaturation cause embrittlement has not yet been firmly established. However, much of the available data supports a mechanism which simply requires that a high internal gas pressure exist at the region of localized high hydrogen concentrations. Such a mechanism is similar to the pressure-expansion theory,¹⁵ except that dislocation movement pumps hydrogen to the voids and thereby generates the high pressures.

The pressure-expansion model for embrittlement is clearly operative in materials that are supercharged either by (1) saturation at elevated temperatures followed by quenching to form a supersaturated solution, or (2) by cathodic charging at very high hydrogen fugacities. In either case, high pressure gaseous hydrogen will precipitate at lattice defects and can cause these defects to grow, coalesce, and lead to blistering and/or specimen failure.

CONCLUSIONS

This discussion attempted to separate hydrogen embrittlement of metals into three categories:

- Embrittlement resulting from blister formation
- Embrittlement resulting from hydride formation
- Generalized embrittlement

Clearly, hydrogen embrittlement is not a simple phenomenon, and many aspects of hydrogen-metal interactions remain unknown. However, within the above classifications, a basis for selection of materials for hydrogen service may be found. Only a small number of metals are hydride formers, and these should not be used in hydrogen atmospheres. Blister formation can be avoided by proper control of process variables, heat treating, and finishing operations. Selection of materials which are resistant to generalized hydrogen embrittlement is more difficult. In general, if an alloy is primarily a transition metal, the alloy should be assumed to be susceptible to embrittlement until proven otherwise. Specific material properties to minimize hydrogen embrittlement are not well established; however, low hydrogen-dislocation binding energies, high stacking-fault energies, low strengths, and low hydrogen solubilities are desirable.

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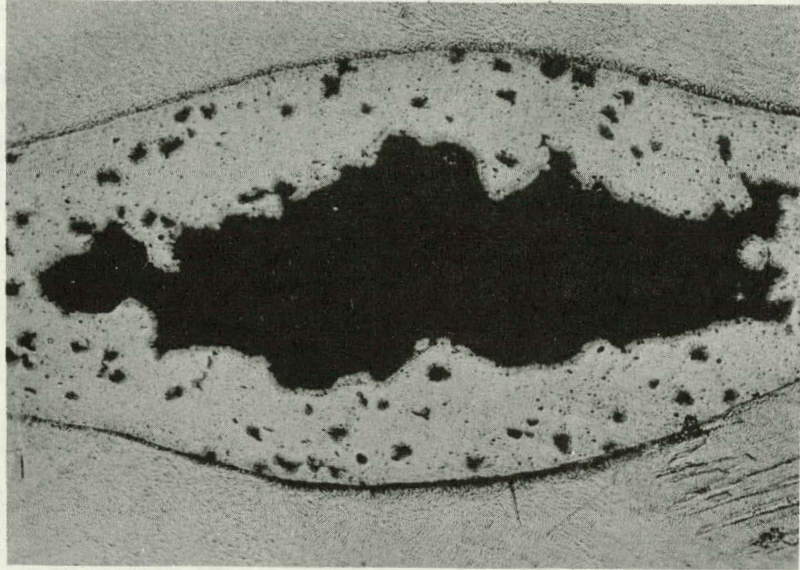


Figure 1a. Hydrogen Blister in Aluminum. Blister formed by exposing 1100 Al to room air in muffle furnace at 773°K.

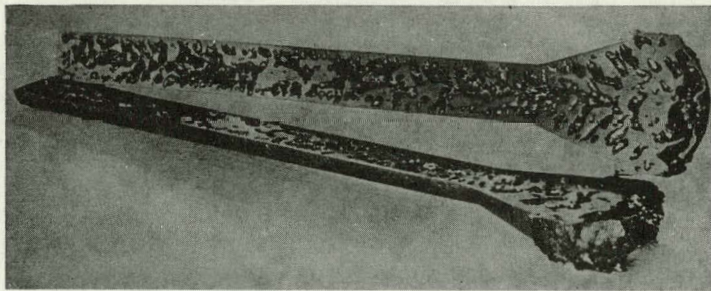
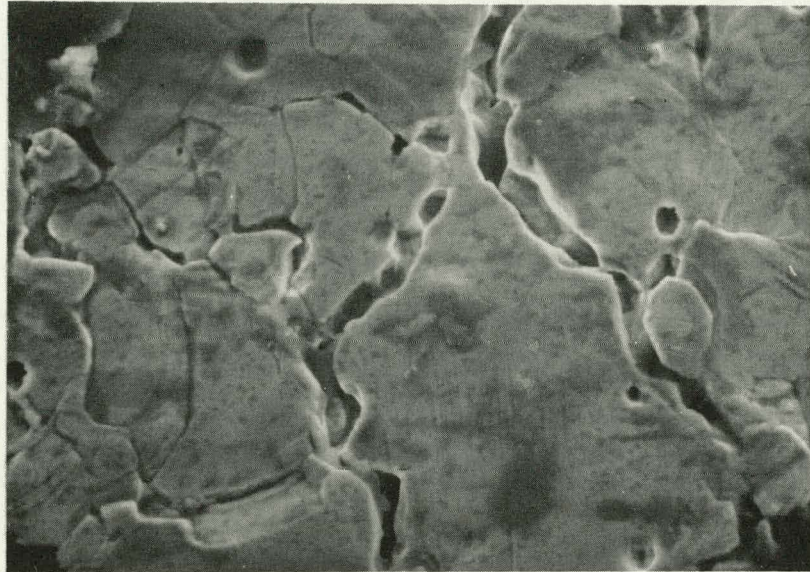
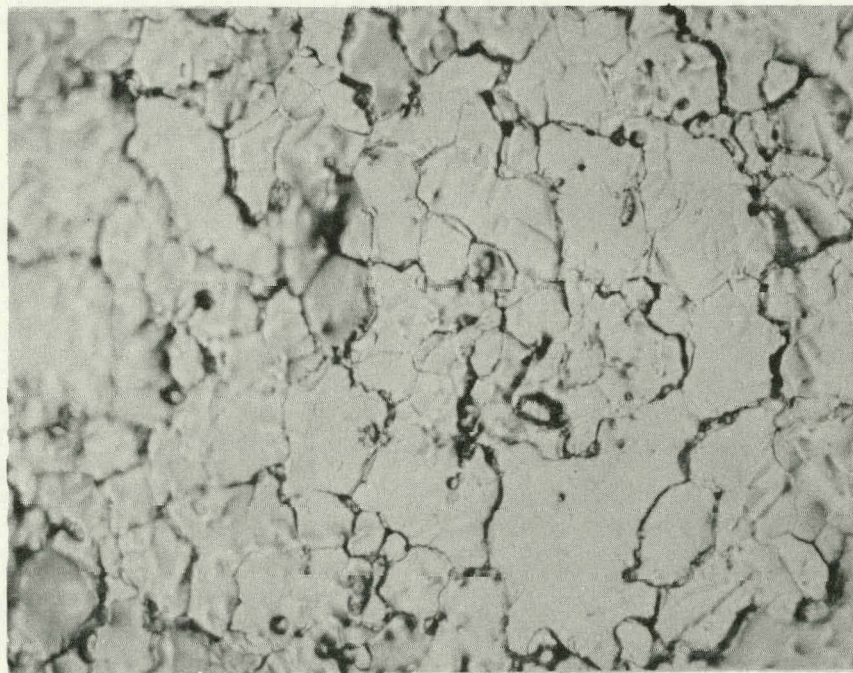


Figure 1b. Hydrogen Blow Holes in Iron Casting. Iron was melted and cast in one atmosphere hydrogen; porosity caused by hydrogen precipitation due to a large drop in solubility upon solidification (from Figure 33 in Reference 6).



10 μm
1.4 cm



20 μm
1 cm

Figure 2. Pores and Cracks along Grain Boundaries of ETP Copper Exposed to Hydrogen at 500°C.

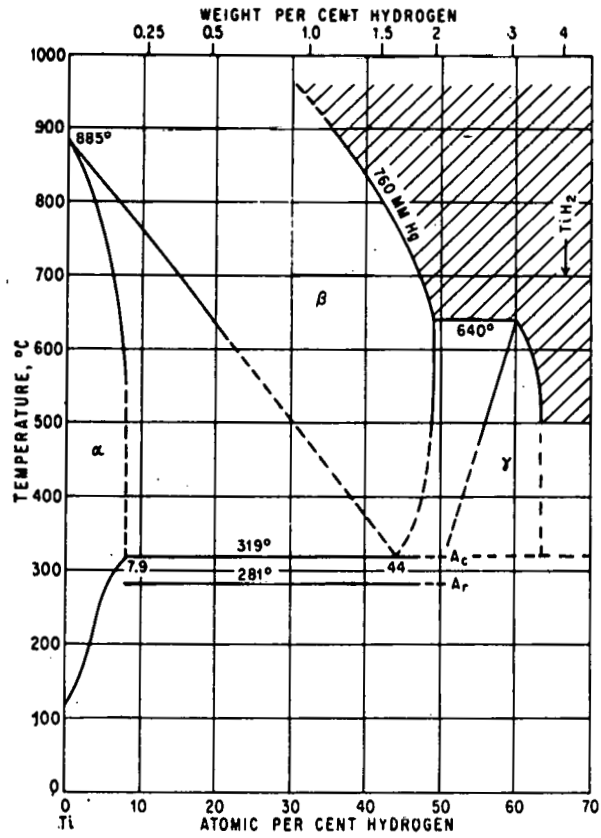


Figure 3. The Titanium - Hydrogen Phase Diagram for Hydrogen at One Atmosphere (from Reference 58).

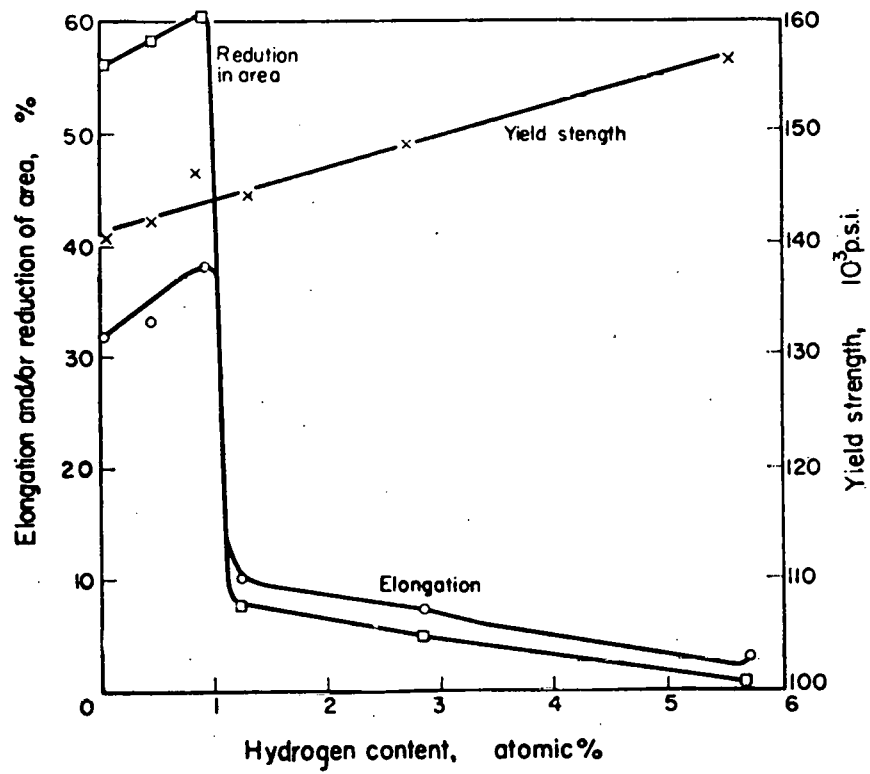


Figure 4. Tensile Properties of Ti-8Mn as a Function of Hydrogen Content.

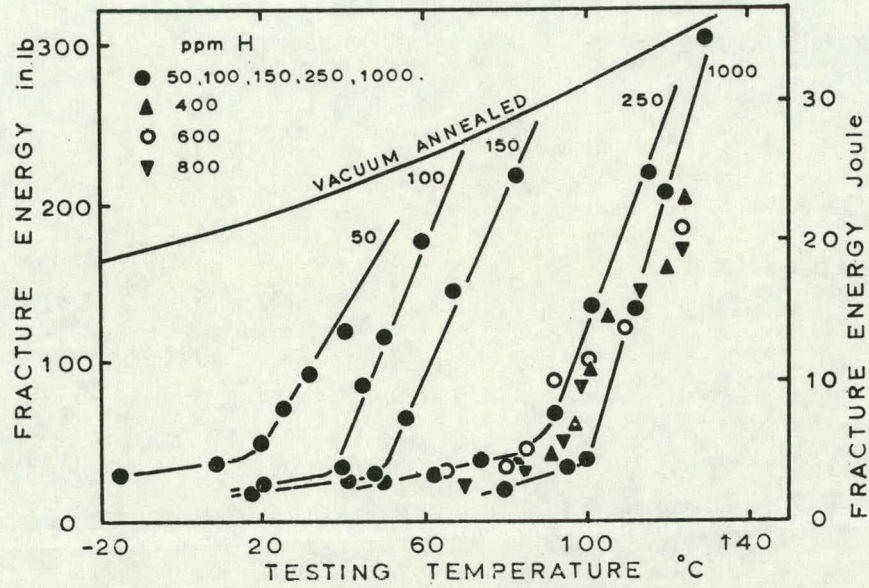


Figure 5. Ductile to Brittle Transitions in Hydrided Zirconium (from Reference 33).

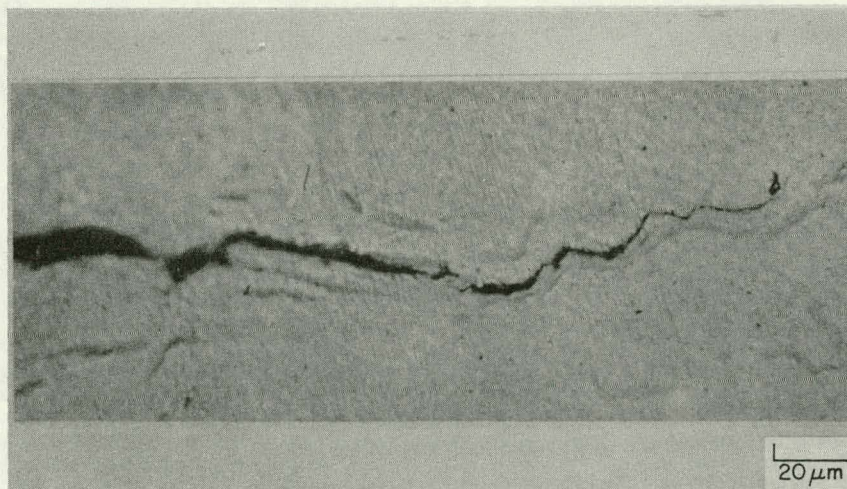


Figure 6. Cracking Along Metal - Hydride Interface. Cracks developed during tensile tests of hydrided Zircoloy-2 (from Reference 59).

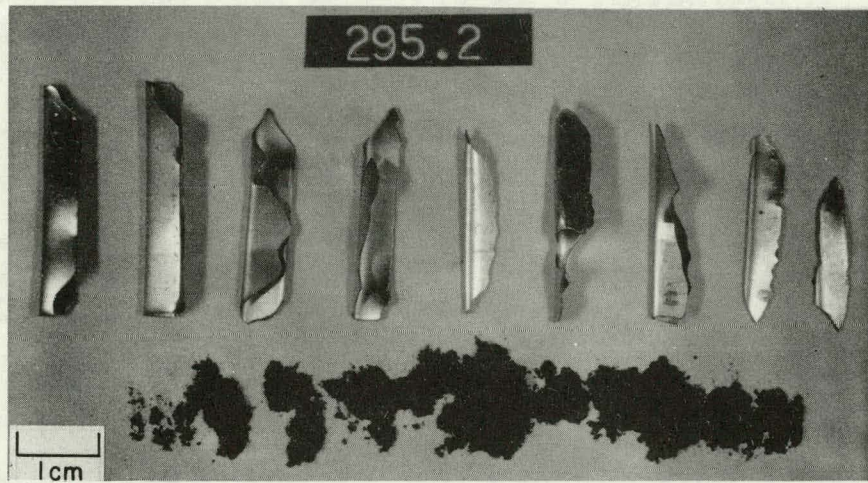
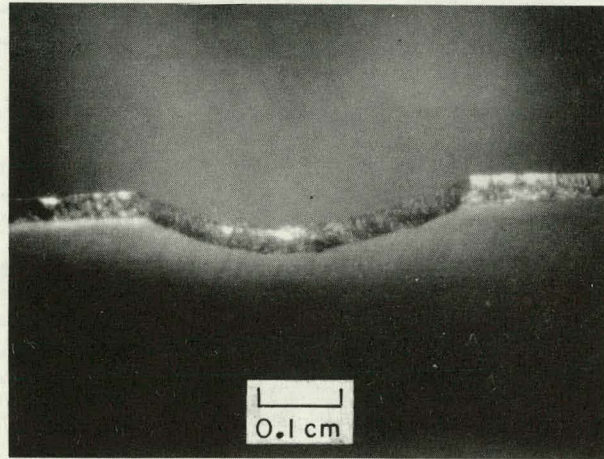


Figure 7. Disintegration of Ti-5Al-2 $\frac{1}{2}$ Sn Specimen Exposed to Hydrogen at 71°C (from Reference 40).

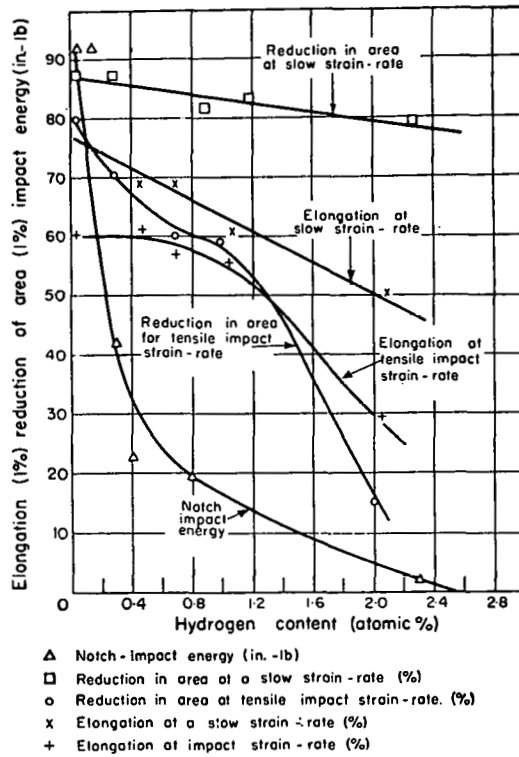


Figure 8. Effect of Testing Techniques in Revealing Susceptibility of Hydrided Titanium to Hydrogen Embrittlement (from Reference 41).

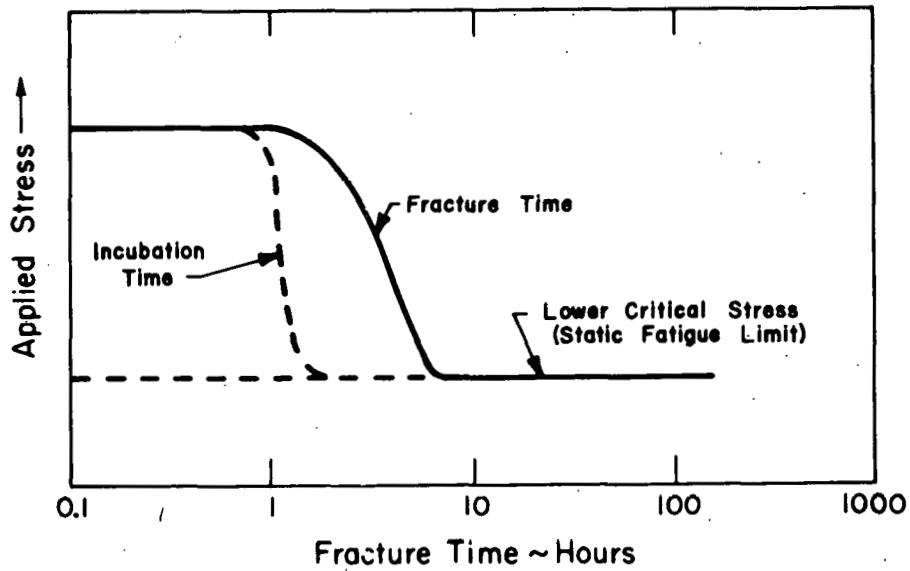


Figure 9. Schematic Representation of Delayed Failure Characteristics of a Hydrogenated Steel (from Reference 44).

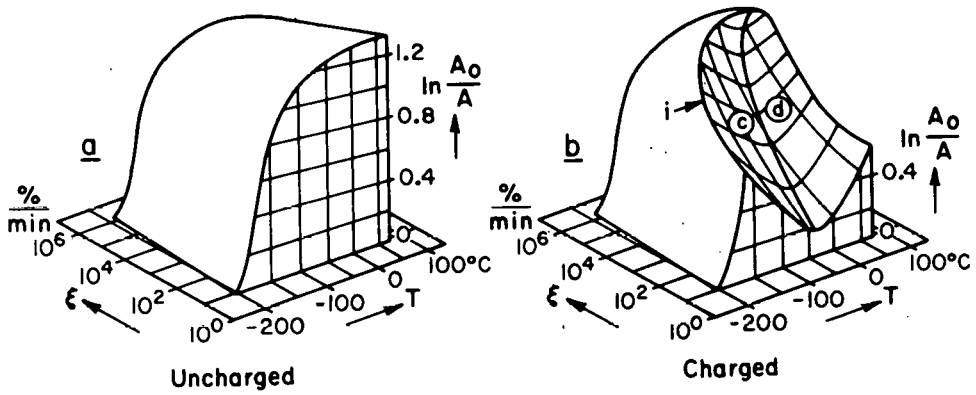


Figure 10 The Effect of Strain Rate and Temperature on the Susceptibility of Mild Steel to Hydrogen Embrittlement (from Reference 46).

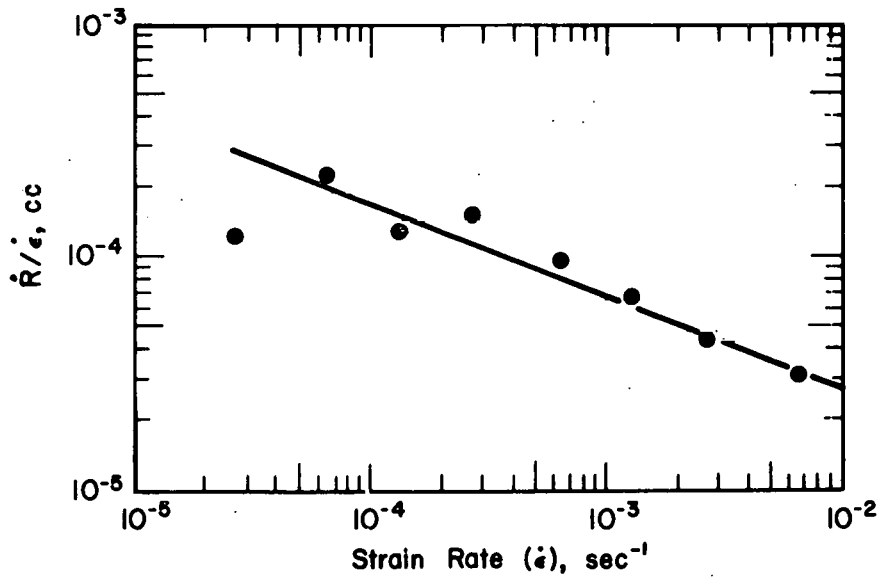


Figure 11. Strain Rate Dependence of Tritium Release from *Armco* Iron Deformed in Tension at 21°C (from Reference 49).

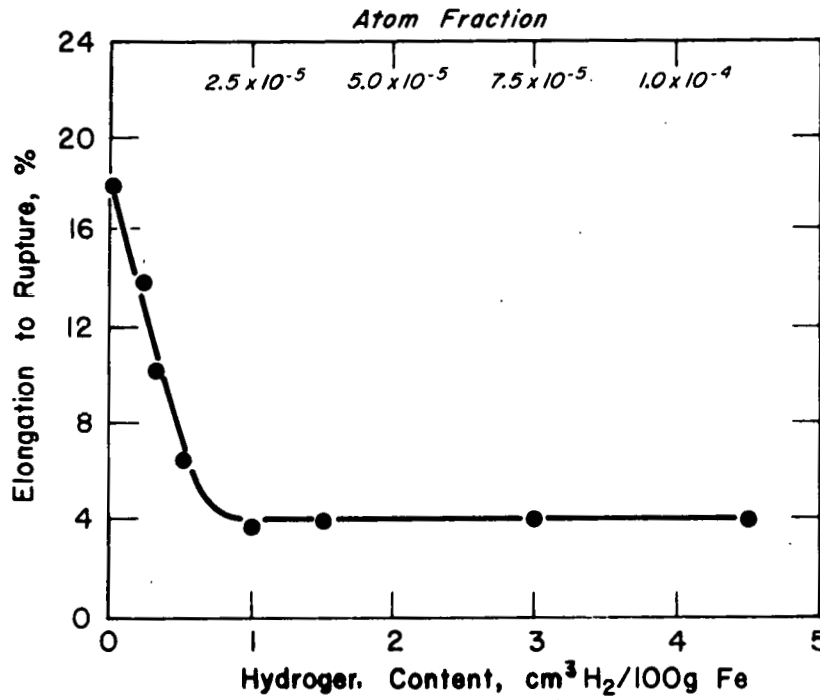


Figure 12. Variation in Elongation to Rupture of Zone Refined Iron with Hydrogen Content (from Reference 47).

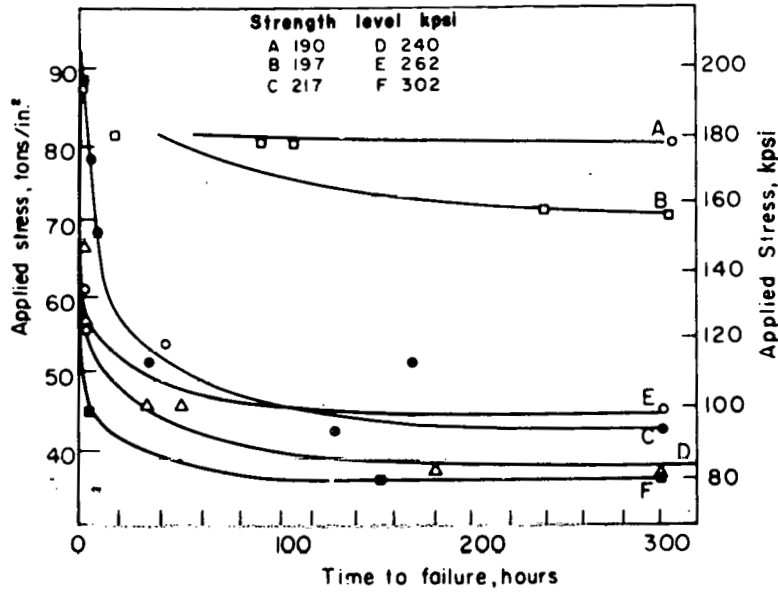


Figure 13. Effect of Strength Level on the Susceptibility to Hydrogen Embrittlement. Data are for static tests; however, similar dependency has been shown⁵ for dynamic tests (from Reference 2).

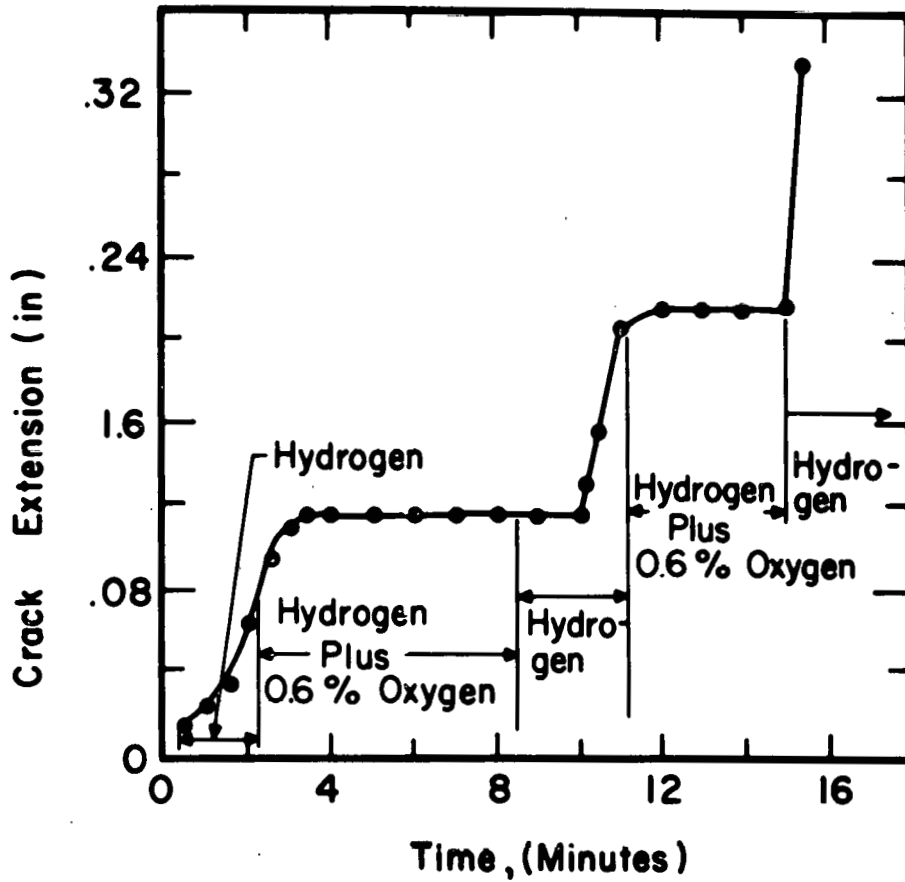


Figure 14. Subcritical Crack Growth in H-11 Steel Exposed to Hydrogen and a Hydrogen -0.6% Oxygen Environment (from Reference 56).