

Atlas of Stress-Corrosion and Corrosion Fatigue Curves

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PREFACE

This Atlas was developed to serve engineers and researchers who are looking for stress corrosion cracking data and/or corrosion fatigue data on particular metal alloys. In the past, the first step to locating this data was an expensive and time consuming search through the technical literature. Now, many of the important and frequently referenced curves are presented together in this volume. The curves are arranged by standard alloy designations and accompanied by a textual explanation of test procedures and interpretation of results. In each case the individual curve is referenced to the original source.

Having these important curves compiled into a single book will also facilitate the computerization of these data. According to current plans, the data presented in this book will soon be available in ASCII format for computer use.

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OTHERS**Mg, Ta, W, Ur, Zn, Zr**

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FERROUS ALLOYS

FERROUS ALLOYS

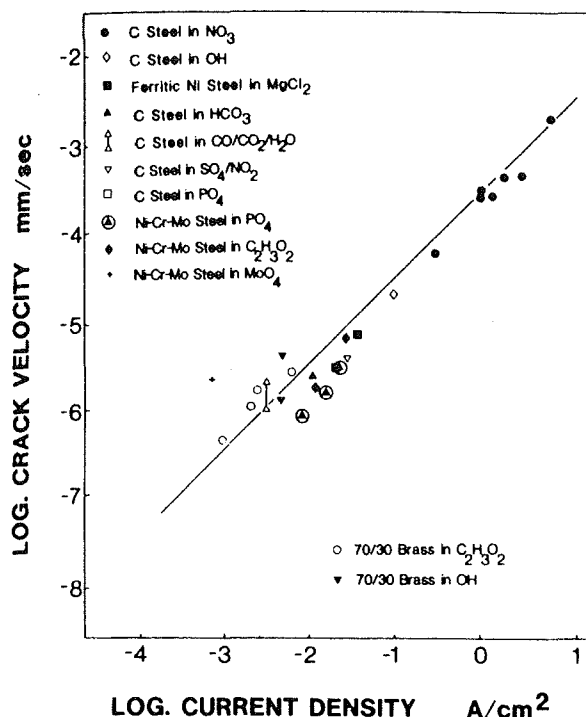
The susceptibility of a steel to stress-corrosion cracking (SCC) is related to its strength level and the nature of the environment in which it is used. For example, a low strength steel exposed to a caustic environment may be subject to stress corrosion cracking due to active path dissolution (APD) processes. Low strength steels are used in boiler construction, and if exposed to a caustic environment, stress corrosion cracking has been known to develop along grain boundaries and lead to explosive failure. The problem was exacerbated in the past by the use of riveted construction which could lead to the development of high misalignment stresses. The presence of crevices between the plates allowed steam, but not the boiler water, to escape, and if the water had been treated with a low concentration of alkali to maintain passivity, the loss of steam would cause the concentration to rise at leakage points and establish the conditions needed for caustic cracking. Fortunately, this problem is no longer as critical. With stress-relieved welded construction, the conditions for cracking are less likely to be present. However, the potential for SCC is still present outside the boiler at the flanged joints.

In high strength steels, hydrogen assisted cracking is more of a concern than in low strength steels, particularly since hydrogen-containing environments are quite common - be they aqueous, gaseous, or produced in processing the steel. Some of the characteristics of hydrogen assisted cracking are shown in the figure in the section on Low Alloy Steels *SA533B-1 High-Strength Low-Alloy Steel: Fatigue Crack Rates in Deoxygenated High-Purity Water*, which indicates the effect of baking at 150°C after charging with hydrogen on stress-lifetime relationships. A critical minimum stress exists below which hydrogen cracking of carbon steel that had been charged with hydrogen will not occur, and this critical stress increases with the decrease in hydrogen content. The cracks themselves appear after a delay time which is a function of hydrogen content and applied stress level.

The potential for corrosion fatigue of ferrous alloys exists in a large number of applications, including offshore structures where salt water is a problem environment, and nuclear reactor pressure vessels where the sulfur content of steels and the oxygen content of the water environment can influence both CF and SCC. As this section indicates, other problem environments exist, such as in oil well drilling operations where the presence of H₂S can worsen conditions for CF as well as SCC.

Maraging steels are subject to SCC in most aqueous environments; their resistance to cracking increases significantly at lower yield strengths. Maraging steels have better resistance to SCC than HSLA steels of comparable strength. Vacuum melting, proper processing, and low levels of impurities improves resistance to SCC.

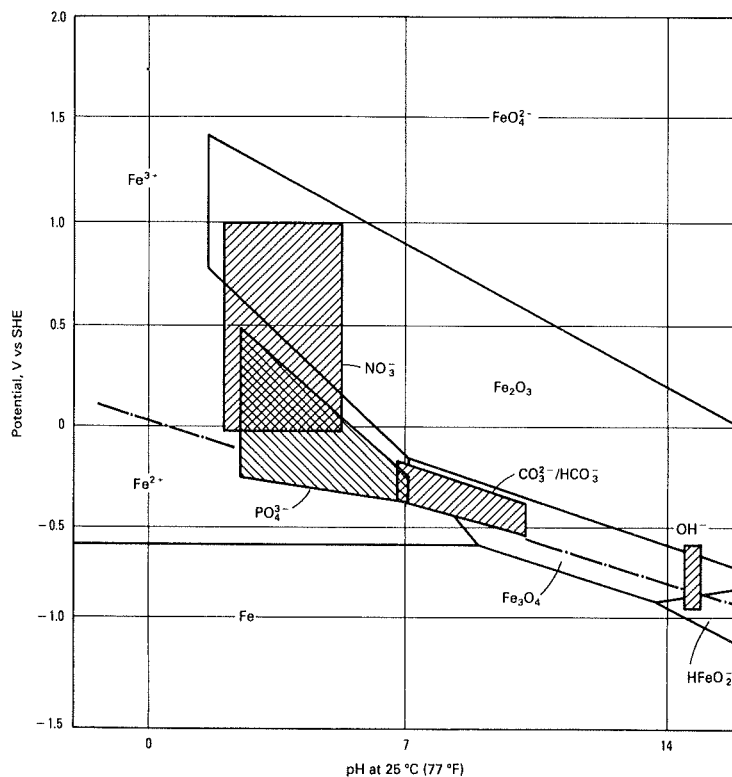
Carbon Steels



Measured stress-corrosion crack velocities and current densities passed at a relatively bare surface for different metals in a variety of solutions.

The line in the above figure is calculated from $da/dt = i_a M / z F d$, where i_a is the anodic current density, M is the atomic weight of the metal, z is the valency of the solvated species, F is Faraday's constant, and d is the density. Such experimental results are obtained from tests on precracked specimens loaded above K_{Isc} or from slow strain rate tests involving relatively high stresses and strain rates. However, as the crack tip strain rate is decreased, the crack velocity reduces by more than an order of magnitude. The results shown above are upper bound results obtained while the tip is bare or unfiled. If the tip strain rate is such that filming may occur, the local crack growth rate will remain zero until the rupture of the film reactivates dissolution. Thus, the processes of metal dissolution, film formation, and film rupture collectively determine the average crack growth rate, which can be anywhere between the upper bound given by the equation and zero.

Carbon Steel: Severe Cracking Susceptibility as a Function of Potential and pH in Various Environments



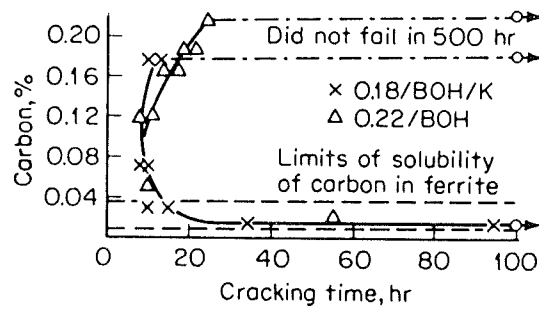
Relationship between pH-potential conditions for severe cracking susceptibility of carbon steels in various environments and the stability regions for solid and dissolved species on the Pourbaix diagram.

Critical potentials for stress-corrosion cracking can be related to potential-pH stability diagrams (Pourbaix diagrams), because these diagrams describe the conditions at which film formation and metal oxidation will occur. The example shown above is for carbon steel in which stress-corrosion cracking is associated with potentials and pHs at which phosphate, carbonate, or magnetite films are thermodynamically stable, whereas species Fe^{2+} and HFeO_2^- are metastable. It is important to note that, as in crevice corrosion, the pH at a growing stress-corrosion crack tip can be much lower than that of the bulk solution. Note that severe susceptibility is encountered where a protective film (phosphate, carbonate, magnetite, etc.) is thermodynamically stable, but if ruptured, a soluble species (Fe^{2+} or HFeO_2^-) is metastable.

For materials in which stress-corrosion cracking occurs by a hydrogen-induced subcritical crack growth mechanism, the thermodynamic requirement for crack growth is governed by hydrogen reduction shown by the dashed line a in the figure. Hydrogen reduction of iron in water at 25 °C occurs at potentials below this line, but not above it. Therefore, the range of potentials at which hydrogen is available to cause crack growth increases and becomes more oxidizing with decreasing pH. Because the pH at a crack tip can be lower than that of the bulk solution, it is quite possible for hydrogen reduction to occur at the tip, although on the basis of the bulk pH value, hydrogen reduction would not be expected.

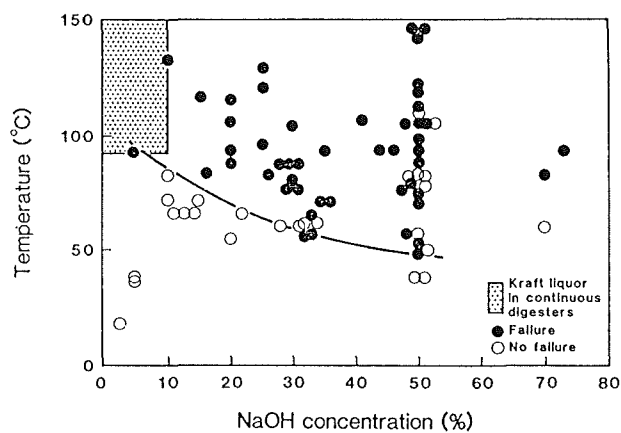
Low-Carbon Steel: Effect of Carbon Content on Cracking Time

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Effect of carbon content on cracking time of low-carbon steel exposed to boiling calcium ammonium nitrate.

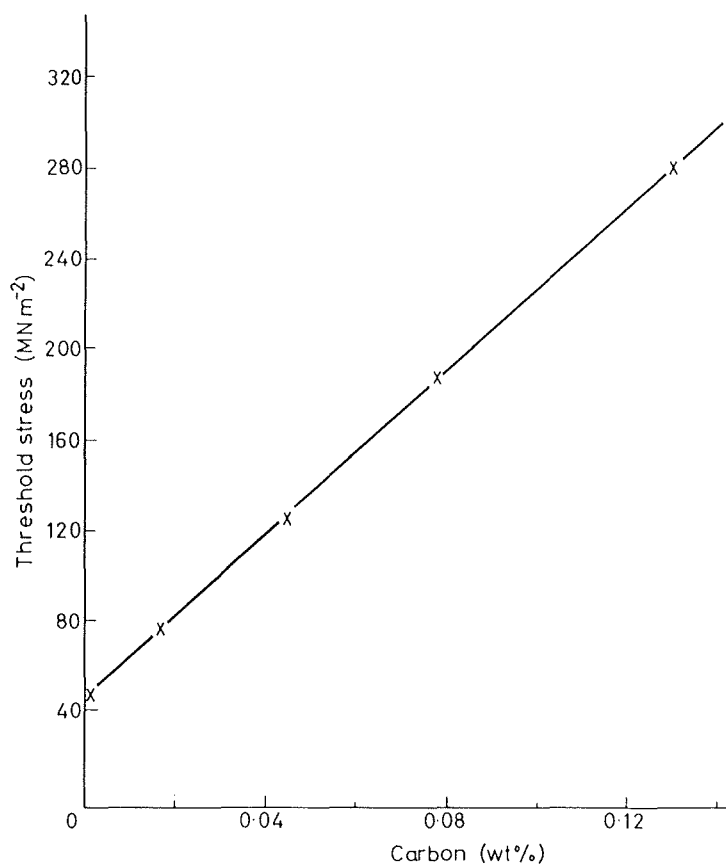
Carbon Steel: Industrial Failures



Results of a 1951 survey of industrial failures of carbon steel vessels resulting from caustic stress-corrosion cracking as a function of NaOH concentration and temperature. Studies have shown that the cracking occurs within a narrow region of electrochemical potentials near the active-passive transition of an anodic polarization curve of carbon steel.

Annealed Low-Carbon Steel: Effect of Carbon Content on Threshold Stress

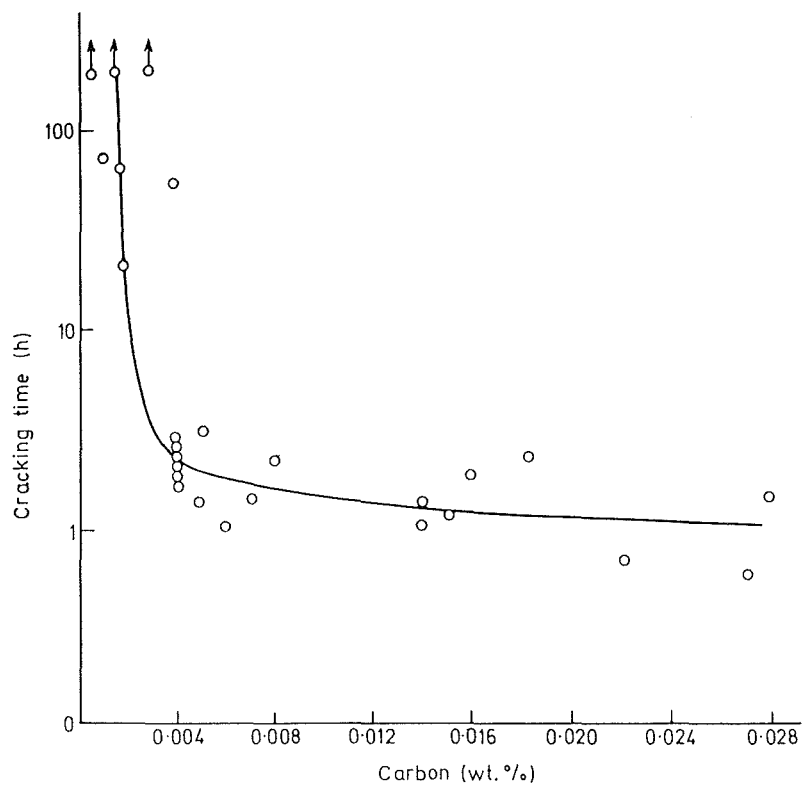
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Effect of carbon content of annealed low-carbon steels on the threshold stress for cracking in boiling 4N NH₄NO₃.

The reduced susceptibility of low-carbon steels to cracking may be due to a change in the distribution of soluble carbon, because an increase in the amount of deformation reduced the amount of carbon detected at the ferrite grain boundaries.

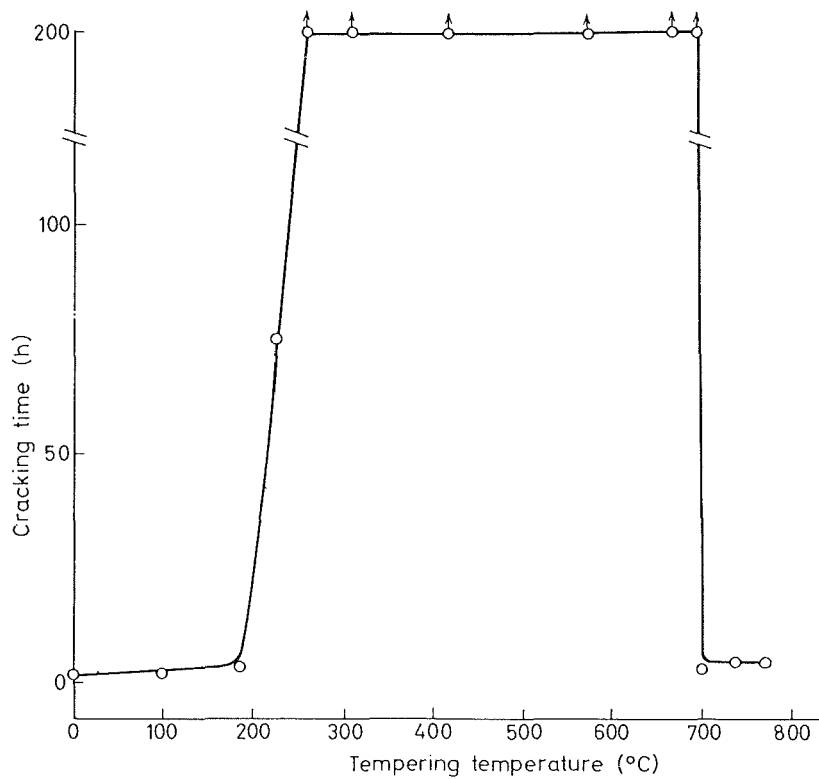
Low-Carbon Steel: Effect of Carbon Content on Cracking



Effect of carbon content of very low carbon steels quenched from 920 °C on cracking in a calcium nitrate/ammonium nitrate solution.

Low-Carbon Steel, 0.06%: Effect of Tempering Temperature/Time

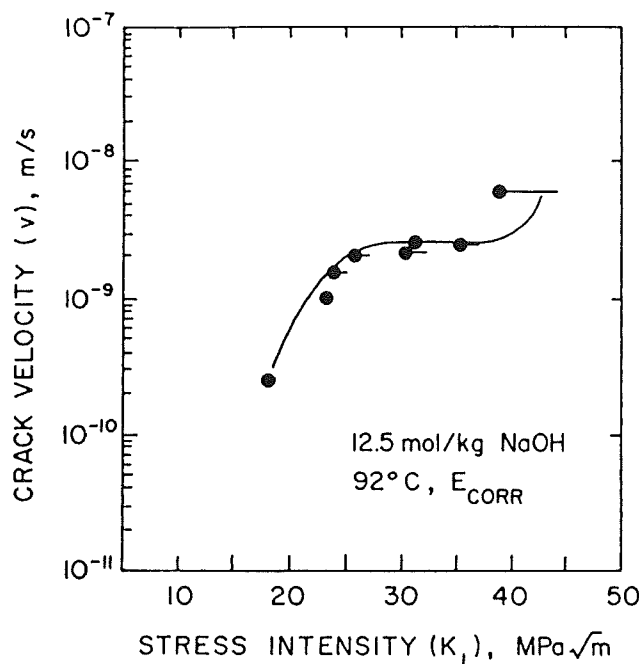
41



Effects of 30 min tempering following quenching from 950 °C on cracking of 0.06% carbon steel in boiling calcium nitrate/ammonium nitrate solution.

In contrast, Houdrement *et al.* found that in a 0.26% carbon steel the increase in susceptibility above 700 °C did not occur. For a low-carbon steel alloyed with 1% Mn, tempering above 200 °C increased susceptibility initially, and increased resistance to cracking began at temperatures above 500 °C. A clear-cut explanation for these various responses is not available.

1018 Low-Carbon Steel: Effect of Stress-Intensity on Crack Propagation Rates

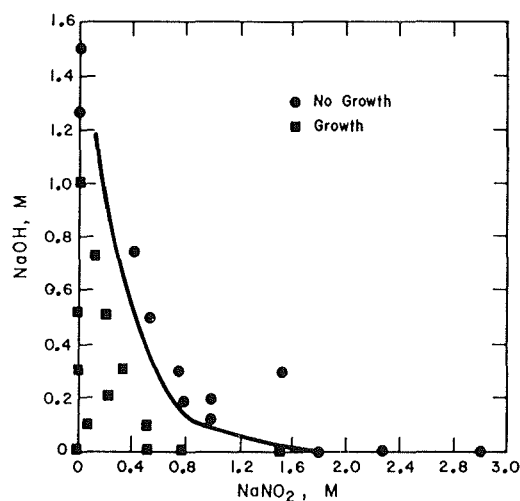


The effect of stress-intensity on crack propagation rates in cold worked 1018 low-carbon steel in aqueous 33% NaOH under freely corroding potentials at 92 °C. $E_{CORR} \approx -1.00 \text{ V}_{SHE}$.

Increasing the potential from E_{CORR} to the active-passive region (-0.76 V_{SHE}) increased the Region II crack growth rate by approximately one order of magnitude. Cracking was transgranular at E_{CORR} and intergranular at -0.76 V_{SHE} , but in neither case was a change in the crack tip pH from that of the bulk pH of 14 detected. It was concluded that the mechanism of cracking was consistent with metal dissolution at -0.76 V_{SHE} and may have involved hydrogen embrittlement and/or embrittlement effects at E_{CORR} . Further studies with a 0.19% carbon steel (SA-516 grade 70) in hot caustic solutions and with the potential in the active-passive region were consistent with a dissolution mechanism of crack advance based on a film rupture-dissolution model.

ASTM A285 B Low-Carbon Steel: Influence of Sodium Nitrite and Sodium Hydroxide on Stress-Corrosion Cracking

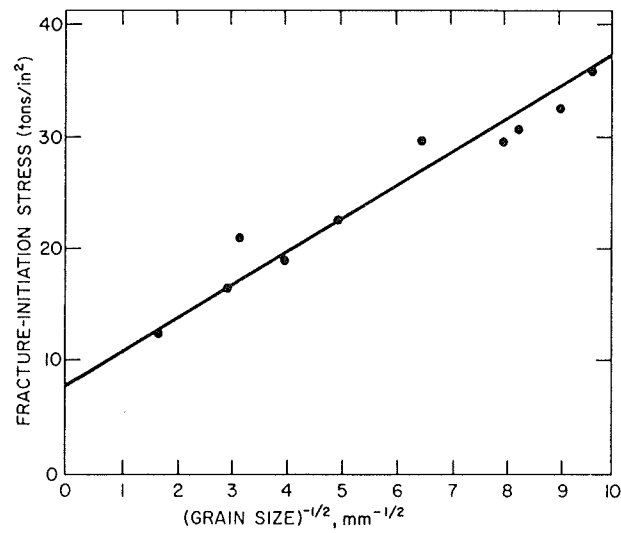
43



Influence of sodium nitrite and sodium hydroxide on stress-corrosion cracking of ASTM A285 B steel in 5M NaNO₃ at 368 K.

Nitrite and hydroxide ions inhibit the growth of nitrate stress-corrosion cracks in low-carbon steel, and because nitrate stress-corrosion cracking involves the dissolution of oxides formed ahead of the crack tip, one role of an inhibitor would be to hinder this attack by making the solution less aggressive toward the oxide film. Another possibility is that nitrite inhibits the cathodic reaction.

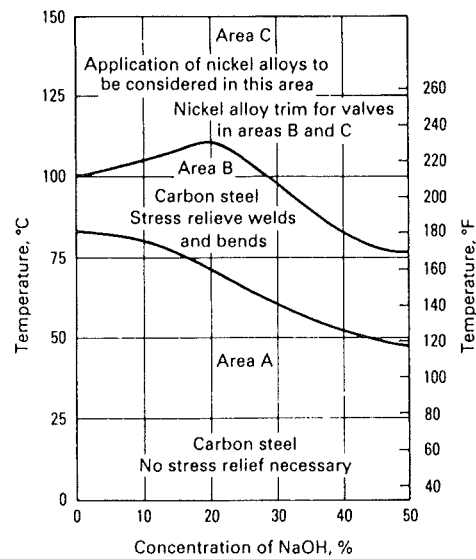
Low-Carbon Steel: Stress-Corrosion Failure as a Function of Grain Size



Stress for the initiation of stress-corrosion failure as a function of grain size in low-carbon steel in a boiling nitrite solution.

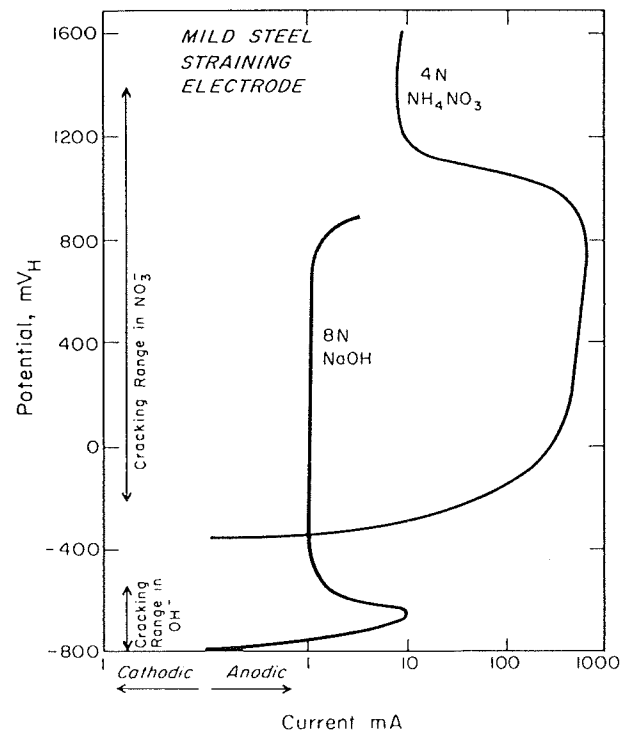
Carbon Steel: Temperature and Concentration Limits for Stress-Corrosion Cracking in NaOH

45



Temperature and concentration limits for stress-corrosion cracking susceptibility of carbon steels in caustic soda.

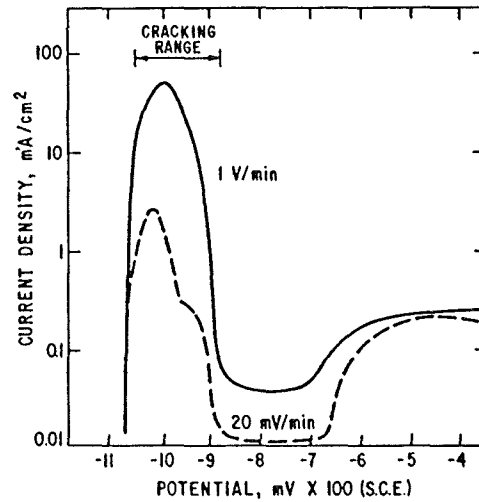
46 **Low-Carbon Steel: Polarization Curves and Potential Ranges, in NaOH and NH_4NO_3 Solutions**



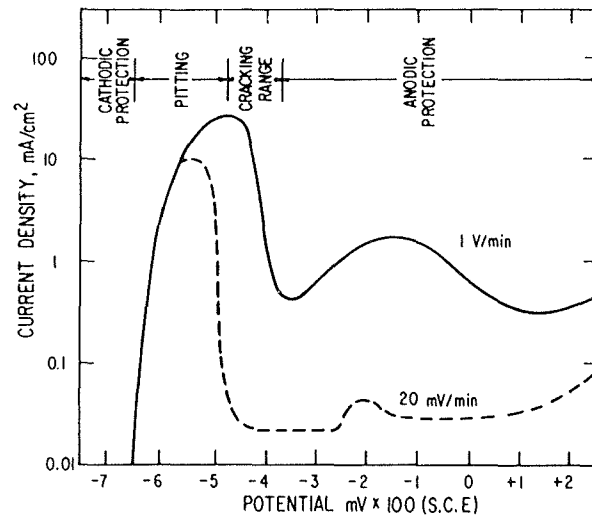
Polarization curves and potential ranges for stress-corrosion cracking for low-carbon steel in caustic and nitrate solutions.

Low-Carbon Steel: Polarization Curves, in NaOH and in $\text{Na}_2\text{CO}_3 + \text{NaHCO}_3$ Solutions

47

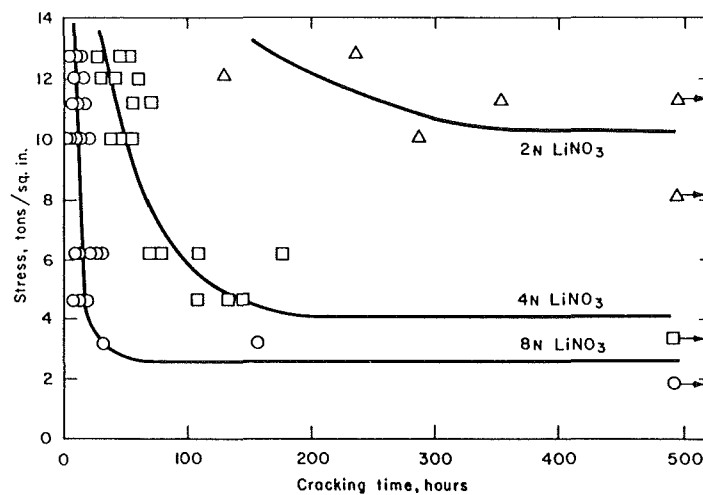


Anodic polarization curves for low-carbon steel in boiling 35% NaOH and the cracking range observed in stress-corrosion cracking experiments. Two different scan rates are shown.



Anodic polarization curves at different potential sweep rates for low-carbon steel in $\text{Na}_2\text{CO}_3 + \text{NaHCO}_3$ at 90°C and the observed results from stress-corrosion cracking tests. Two different scan rates are shown.

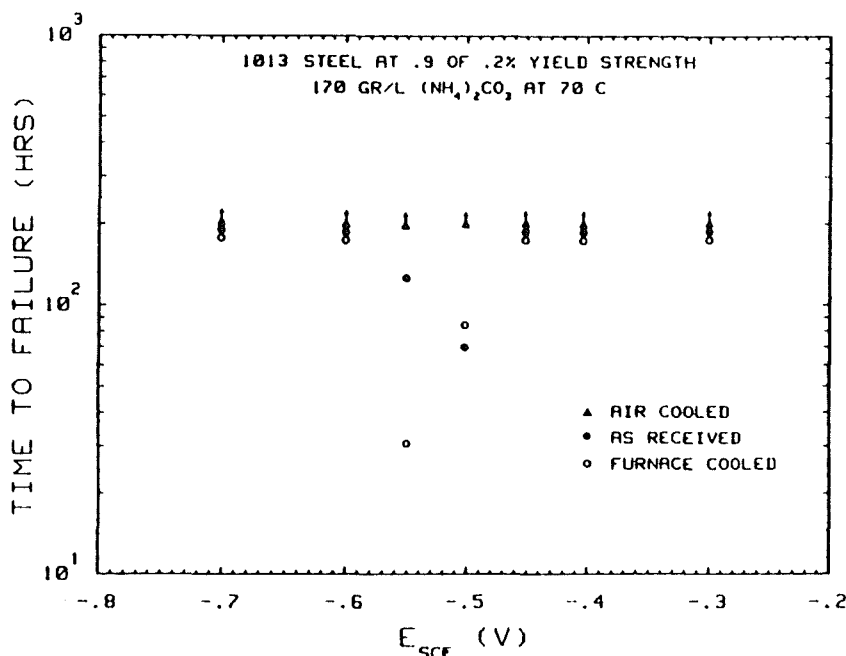
Low-Carbon Steel: Stress-Corrosion Cracking Times



Stress-corrosion cracking times for a low-carbon steel immersed in boiling LiNO_3 solutions.

SAE 1013 Low-Carbon Steel: Time-to-Failure versus Potential in $(\text{NH}_4)_2\text{CO}_3$

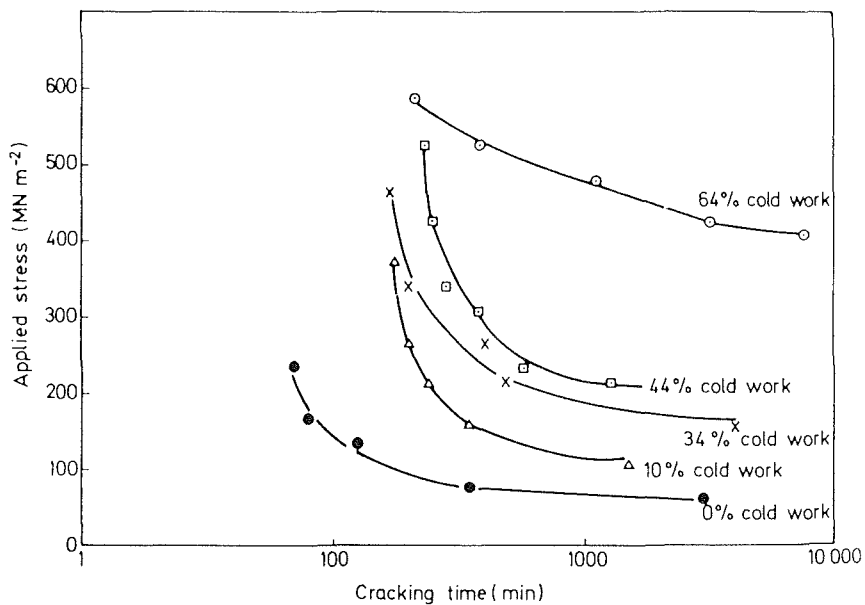
49



Time-to-failure versus potential for 1013 steel in $(\text{NH}_4)_2\text{CO}_3$ at 70 °C. The applied stress was 90% of the yield stress.

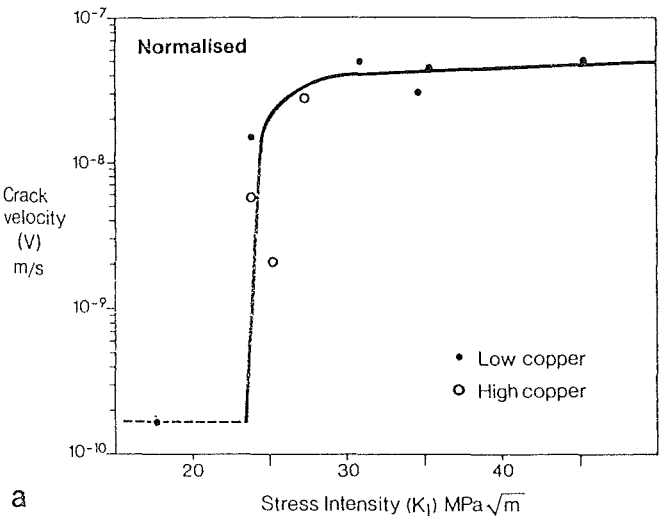
In contrast to results reported for nitrate solutions, in carbonate solutions, slow cooling from an austenitizing temperature induces intergranular stress-corrosion cracking susceptibility, whereas rapid cooling decreases susceptibility. It was proposed that the specific chemistry of the grain boundaries allows different passivity between the bulk alloy and the boundaries and that the role of mechanical strain is to continuously rupture the passive film at the crack tip.

Low-Carbon Steel, 0.017%: Effect of Plastic Strain on Cracking Susceptibility, in NH_4NO_3

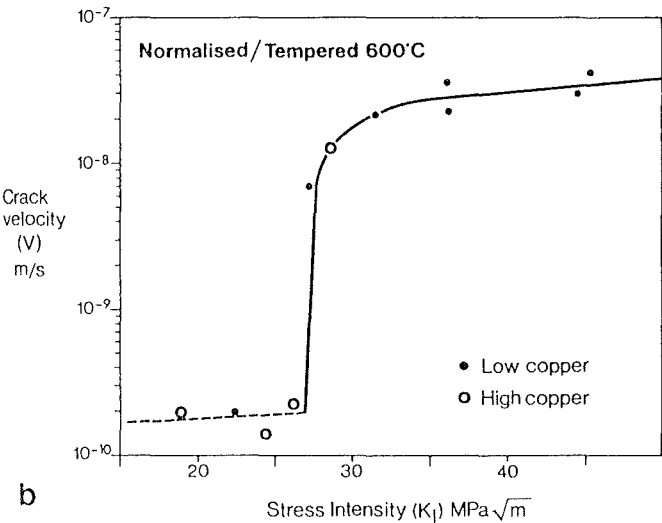


The effects of various amounts of prior plastic strain on the cracking susceptibility of an 0.017% carbon steel in boiling 4N NH_4NO_3 .

Low-Carbon Steel: Stress-Corrosion Crack Velocity as a Function of Stress-Intensity

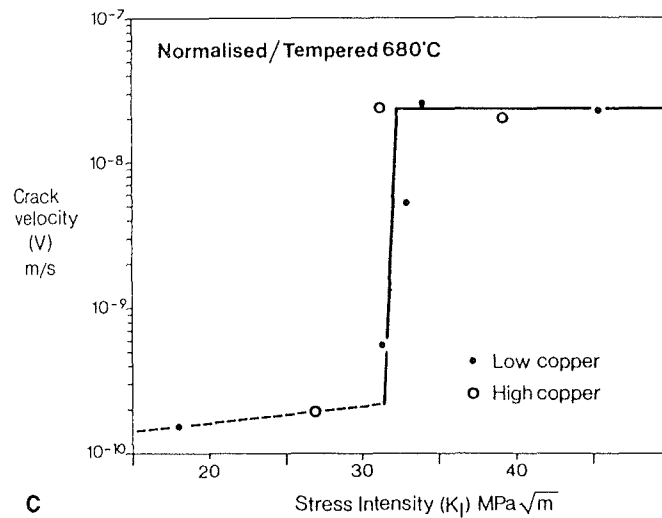


a



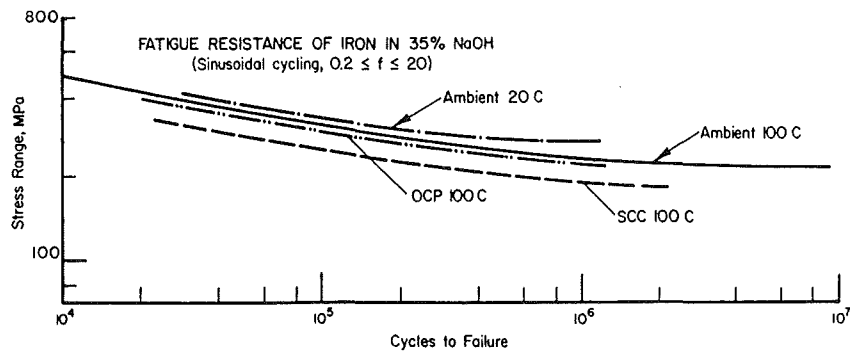
b

Low-Carbon Steel: Stress-Corrosion Crack Velocity as a Function of Stress-Intensity (Continued)



Stress-corrosion crack velocity (V) as a function of applied stress-intensity (K_I). (a) Normalized low-carbon steel, (b) normalized low-carbon steel tempered for 1 h at 600 °C, and (c) normalized steel tempered at 680 °C.

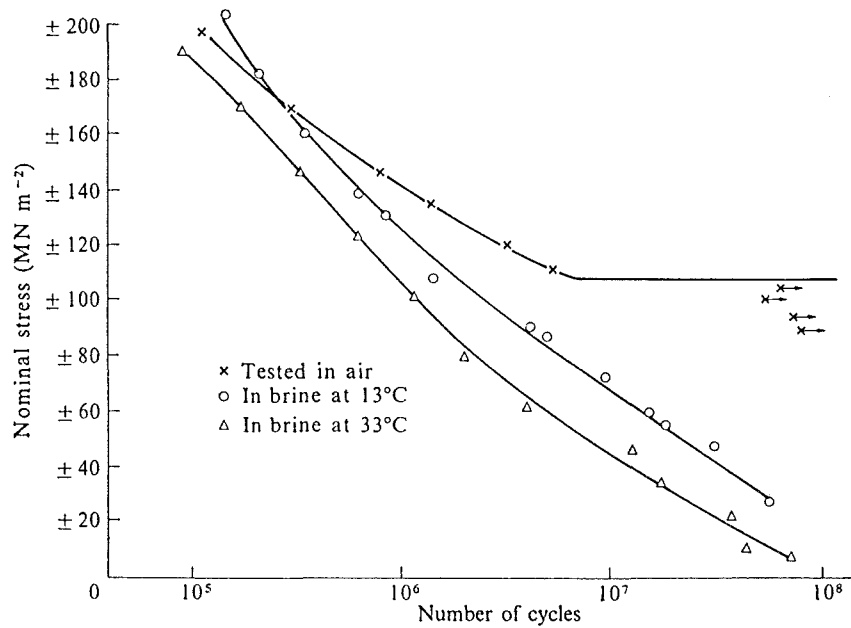
Stress-corrosion cracking velocity measurements were obtained using zero stress-intensity gradient double-cantilever beam specimens. It was proposed that at low stress intensities the buildup of corrosion products on the crack surfaces closes off the electrochemical processes, and crack arrest occurs.



Observed environmentally assisted fatigue behavior of Armco iron cycled in displacement control.

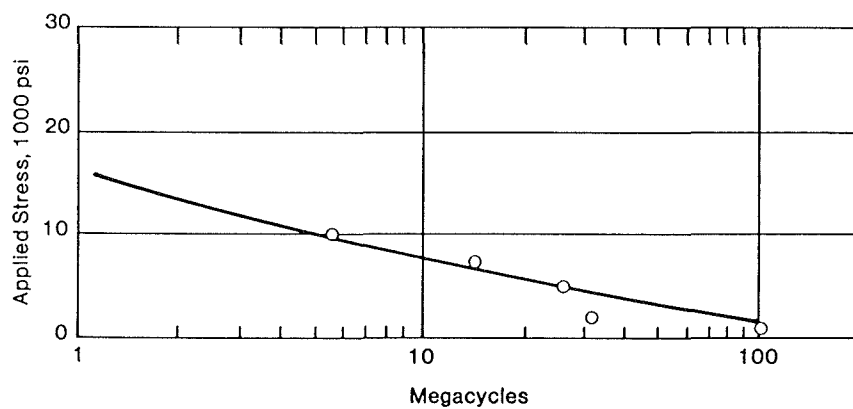
Source: B.N. Leis, Crack Initiation Due to Cyclic Loading in an Aggressive Environment, in Aspects of Fracture Mechanics in Pressure Vessels and Piping, S.S. Palusamy and S.G. Sampath, Ed., American Society of Mechanical Engineers, New York, 1982, p 313-322.

Low-Carbon Steel: S/N Curves for V-Notched Specimens, in Air and in Brine



S/N curves for V-notched low-carbon steel specimens tested in brine.

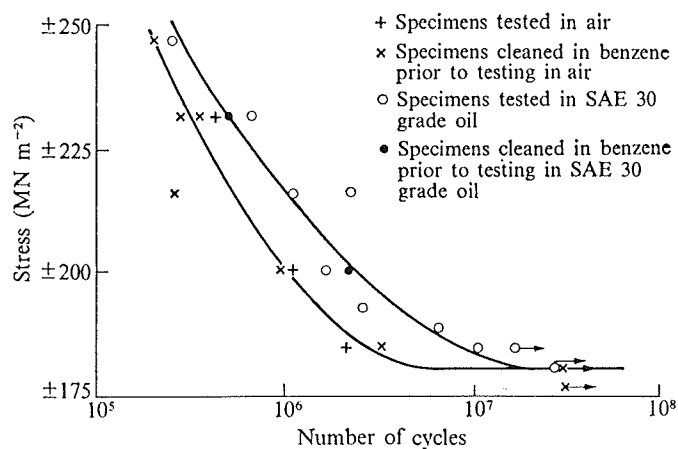
The specimens contained a circumferential V-groove, 2.5 mm deep, and 0.05 mm root radius. In brine tests, cracks form rapidly, and the lifetimes represent resistance to crack growth.



Corrosion fatigue of low-carbon steel rotated at 1400 rpm in seawater.

Note that at 100 megacycles the fatigue strength has been virtually reduced to zero.

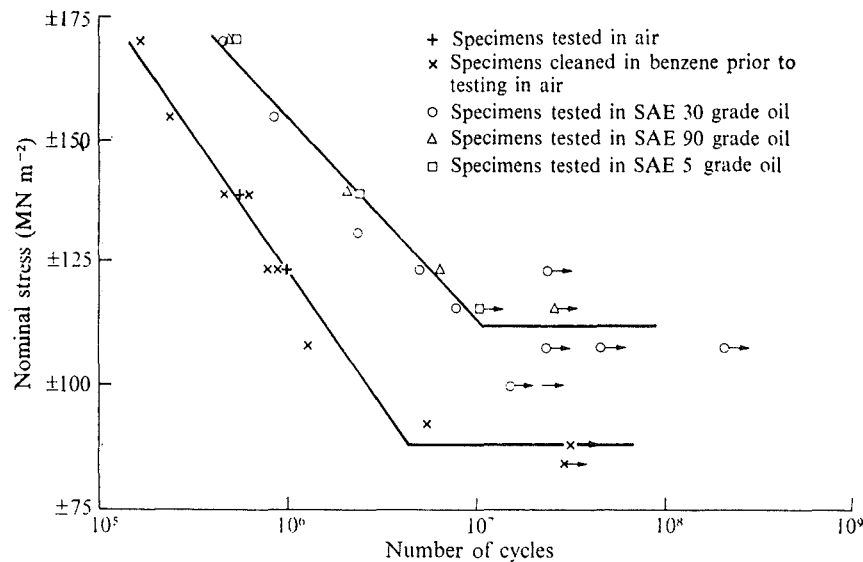
Low-Carbon Steel: S/N Curves for Rotating Bending, in Air and in Oil



S/N curves for low-carbon steel rotating bending plain specimens tested in air and in SAE 30 grade oil.

Note that there is little effect of the environment at long lives, but that specimens tested in oil have longer lives in the finite life range, probably due to the exclusion of moisture.

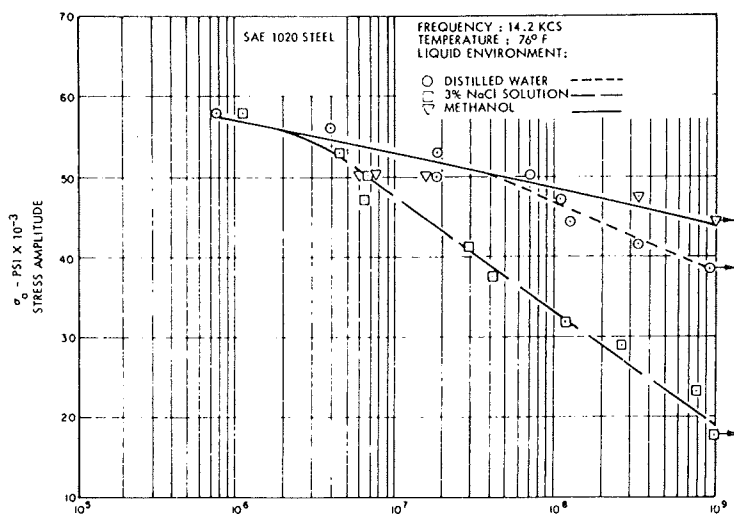
Low-Carbon Steel: S/N Curves for Sharply Notched Specimens, in Air and in Oil



S/N curves for sharply notched low-carbon steel specimens tested either in air or immersed in mineral oil.

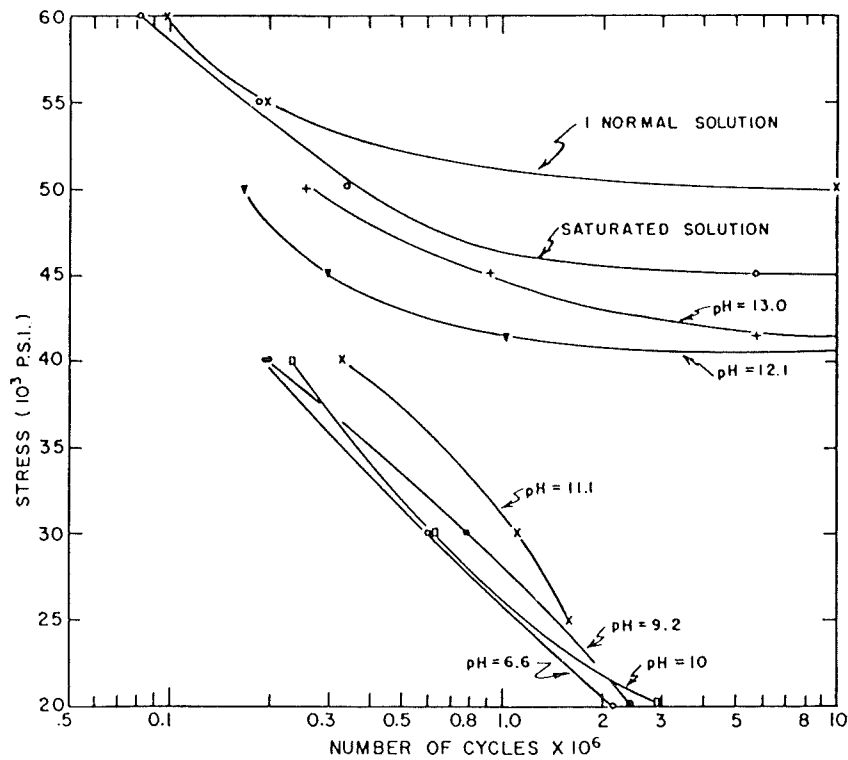
In contrast to the smooth specimen behavior (26, for sharply notched specimens), both the finite life and infinite life regions are improved when moisture is excluded. For the notched specimens, cracks are more easily nucleated than in smooth specimens, and the effect of the environment on the endurance limit reflects a similar dependency of the threshold level for fatigue crack propagation on the environment.

1020 Low-Carbon Steel: High-Frequency Corrosion Fatigue, in Water, NaCl, and in Methanol



High-frequency corrosion fatigue of SAE 1020 steel.

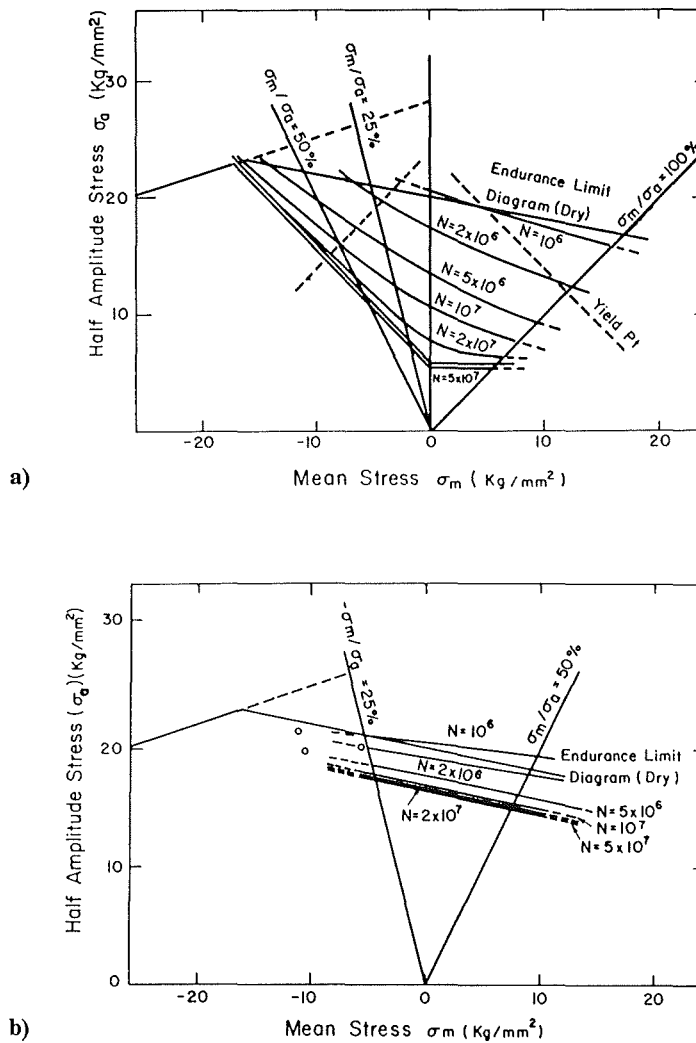
Note that the high frequency of testing allowed the number of cycles to reach 10^9 cycles and that the fatigue strength continues to decrease even at this lifetime.



The effect of pH on the fatigue behavior of low-carbon steel in NaCl + NaOH.

Note that a fatigue limit is observed at pH 12.1. The authors suggested that corrosion fatigue resulted from development of corrosion pits on the specimen surfaces due to the presence of differential aeration cells. At pH levels above 12.1, the presence of a fatigue limit was attributed to a more protective film barrier.

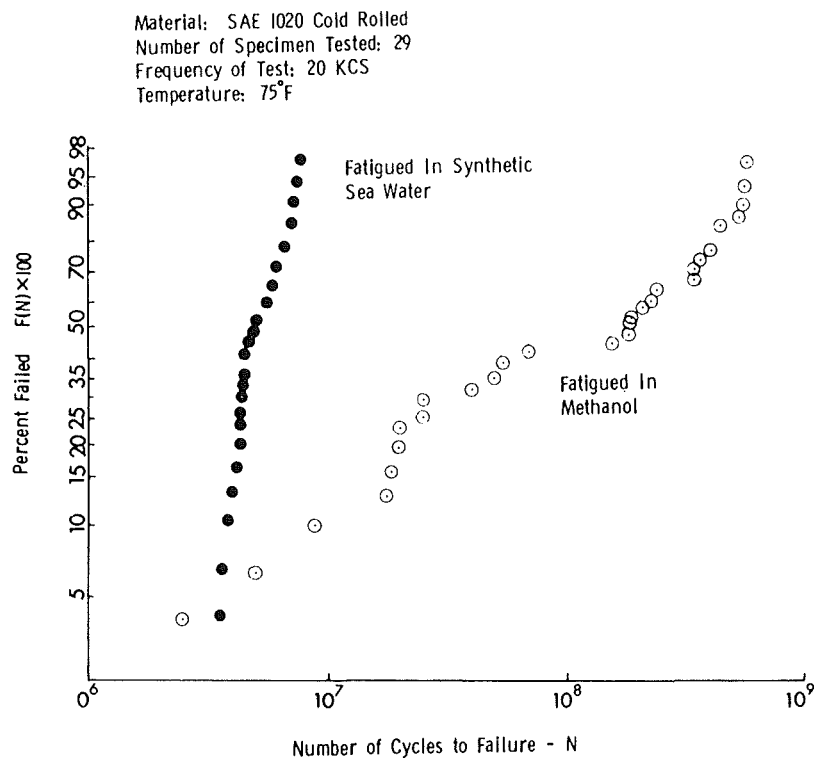
Low-Carbon Steel: Effect of Mean Stress on Corrosion Fatigue Lifetime, in NaCl and in Water



The effect of mean stress on the corrosion fatigue lifetime. (a) Corrosion fatigue strength of low-carbon steel in NaCl solution for a specified number of cycles, N . (b) Corrosion fatigue strength of low-carbon steel in moderately fresh water for a specified number of cycles, N .

1020 Low-Carbon Steel: Weibull Distribution of High-Frequency Fatigue, in Methanol and Seawater

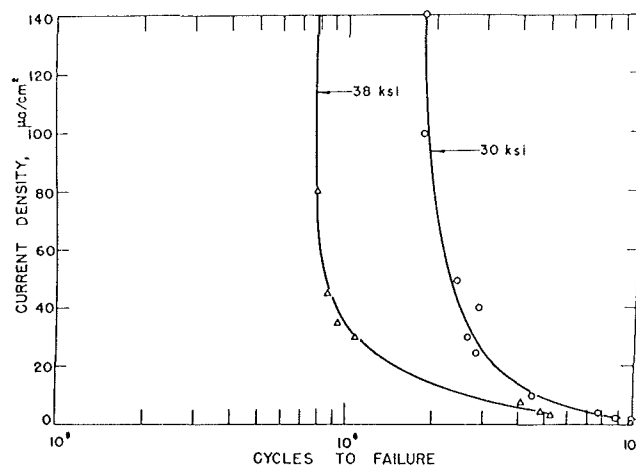
61



Weibull distribution of high-frequency (20 kHz) fatigue lifetimes for cold rolled SAE 1020 steel.

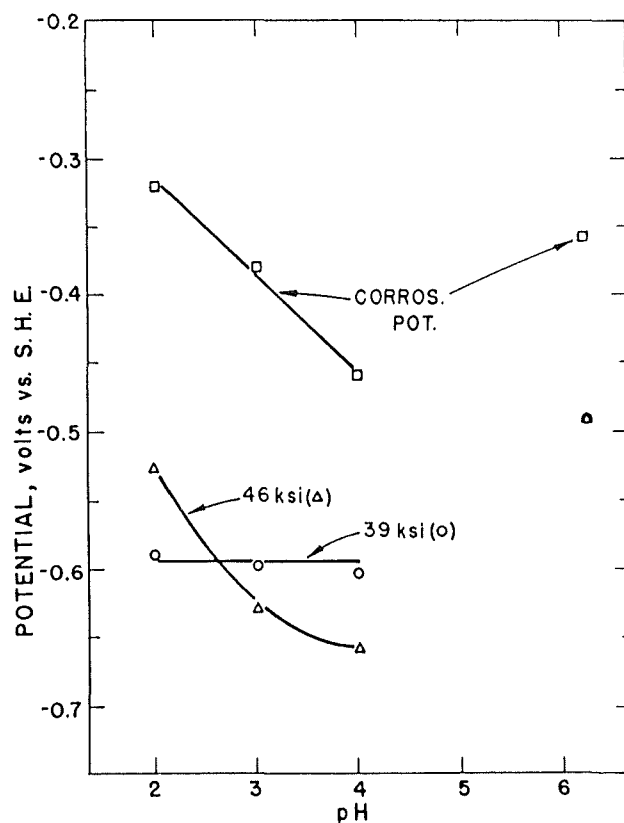
The steeper slope in seawater reflects a lesser degree of scatter than in methanol. Lack of scatter results from early crack initiation with the lifetime being controlled by crack propagation. There is less scatter in crack propagation, because the properties of many grains are being averaged along the crack front, as compared to crack initiation which is sensitive to the weakest regions in the surface.

Low-Carbon Steel: Effect of Applied and Anodic Currents on Fatigue Life



The effect of applied and anodic currents on the fatigue lives of low-carbon steel in deaerated 3% NaCl solution.

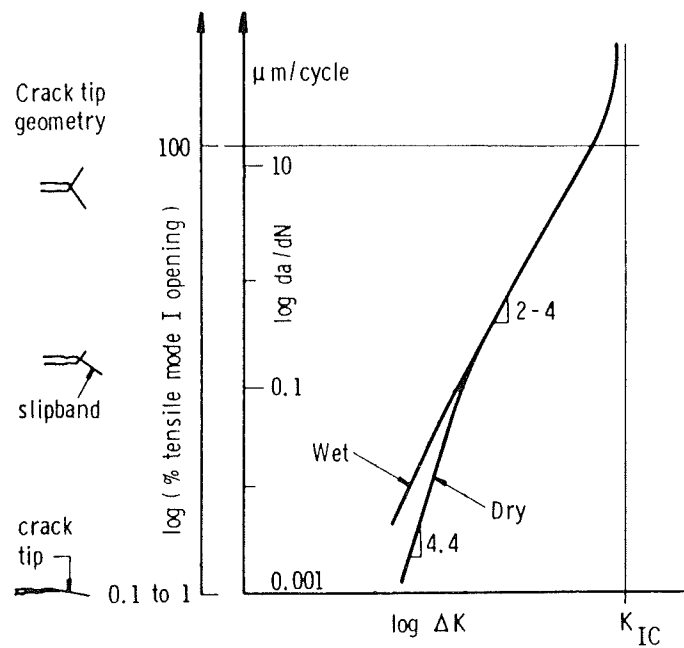
The corrosion rate of this steel in this solution is virtually zero in the absence of applied currents. Note the independence of fatigue life at currents greater than $\sim 40 \mu\text{A}/\text{cm}^2$ and at currents less than $\sim 0.2 \mu\text{A}/\text{cm}^2$ the absence of an applied stress effect and the reappearance of the fatigue limit. These observations indicate that a critical corrosion rate is necessary to initiate corrosion fatigue failures.



Corrosion potentials and corrosion-fatigue potentials for low-carbon steel in NaCl + HCl as a function of pH.

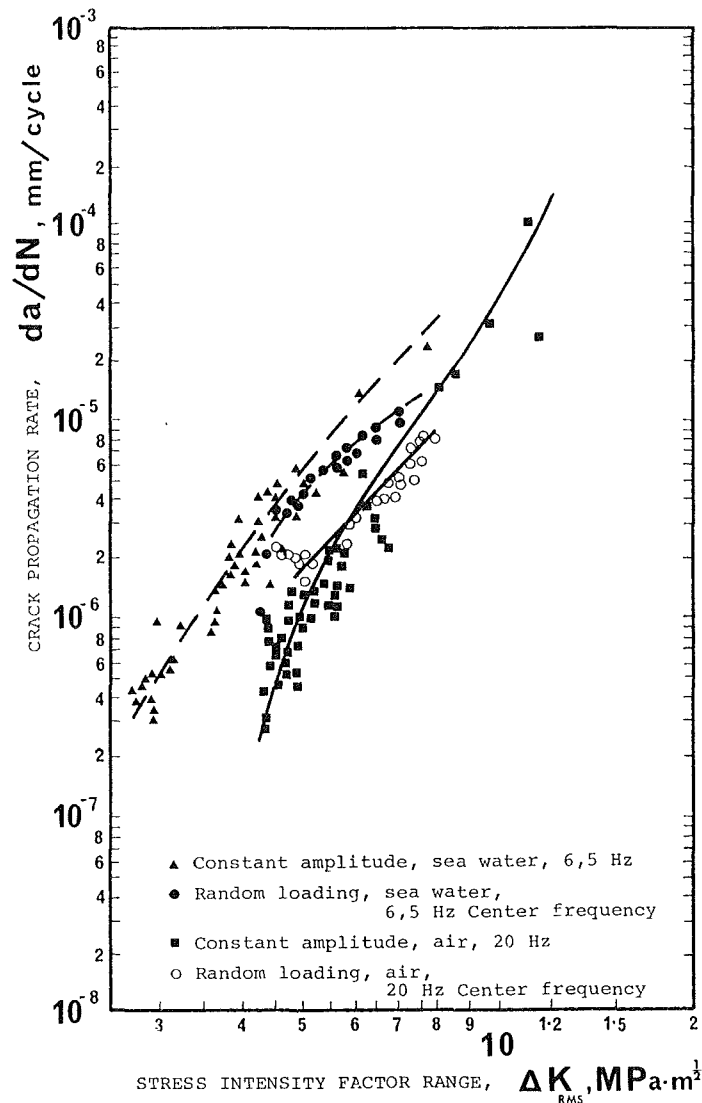
The fatigue limit in air is between 39 and 46 ksi. Note that the protection potential is independent of pH below the fatigue limit in air, but not above the fatigue limit. Below the fatigue limit, the critical potential for complete protection is nearly identical to the calculated potential of an iron surface in equilibrium with ferrous ion in solution.

Low-Carbon Steel: Factors Affecting Crack Growth When Exposed to Wet Air and Dry Nitrogen



Schematic relationship among crack opening mode, extent of plastic zone, and environmental effects in crack growth in low-carbon steel exposed to wet air and dry nitrogen.

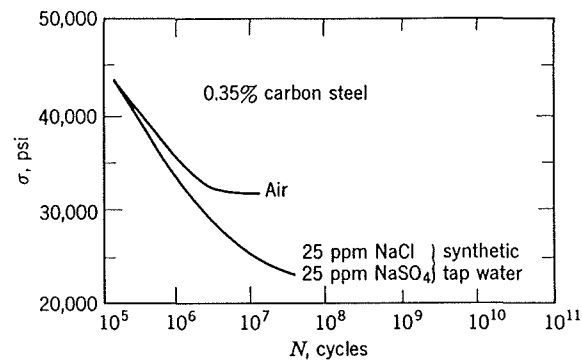
Note that the increase in crack propagation rate in wet air was observed only at low ΔK levels.



Effect of environment on the fatigue crack propagation rates in a low-carbon steel as a function of the range of the stress-intensity factor for both constant-amplitude and random-amplitude loading.

Note that the random-amplitude test results correlate with the constant-amplitude test results when an rms value of the random loading is used.

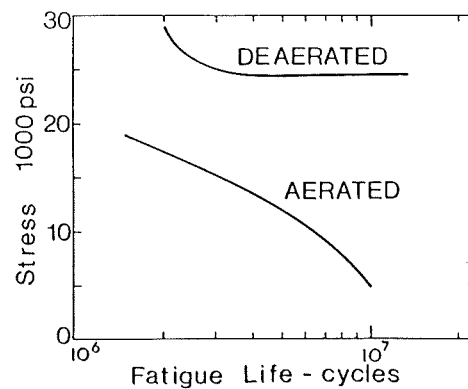
Medium-Carbon Steel, 0.35%: Effect of Environment on Fatigue Properties



Effect of environment on fatigue properties of a 0.35% carbon steel. Note the absence of an endurance limit under corrosion fatigue conditions. In some instances, the corrosion fatigue resistance at long lives can sharply decrease if corrosion pits develop late in the fatigue lifetime.

1035 Medium-Carbon Steel: Effect of Dissolved Oxygen on Fatigue Behavior, in NaCl Solution

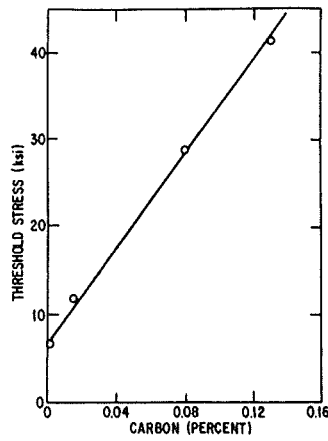
67



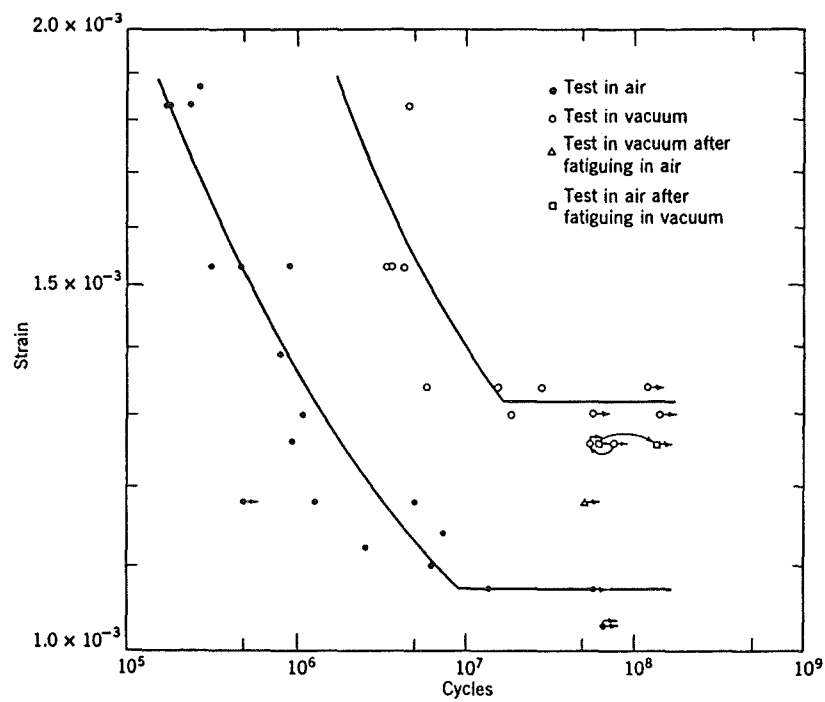
The effect of dissolved oxygen on the fatigue behavior of 1035 steel in 5% NaCl solution.

A fatigue limit is present in the completely deaerated sodium chloride solution, an indication that dissolved oxygen plays an important role in the corrosion fatigue process of carbon steels in neutral solutions.

Fe-C Carbon Steel: Effect of Carbon Content on Threshold Stress for Stress-Corrosion Cracking, in Nitrate Solutions



Effect of carbon content on threshold stress for stress-corrosion cracking of iron-carbon alloys in nitrate solutions.

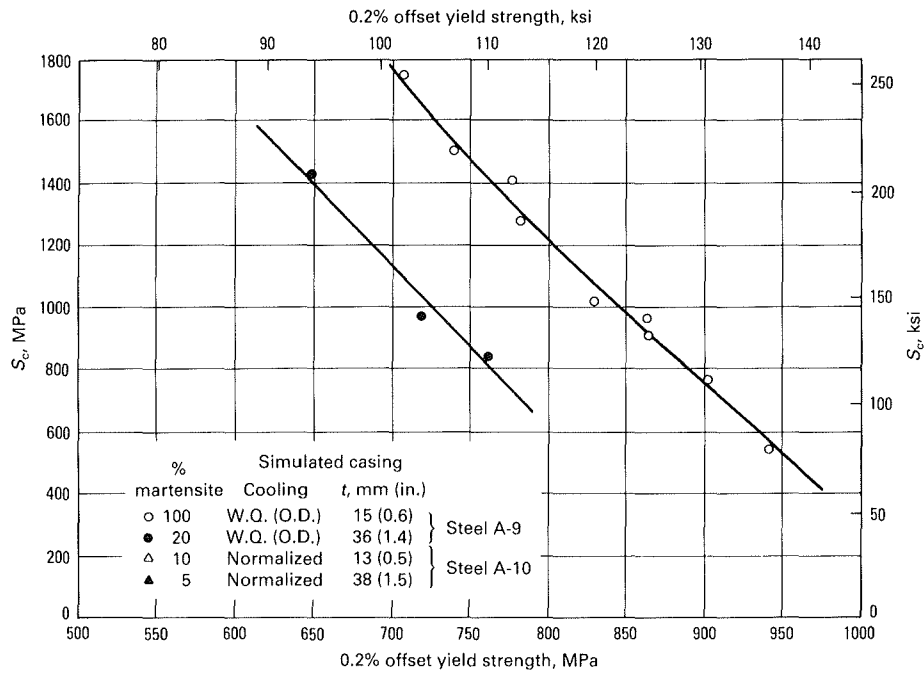


Effect of environment on fatigue properties of an Fe-0.5C steel.

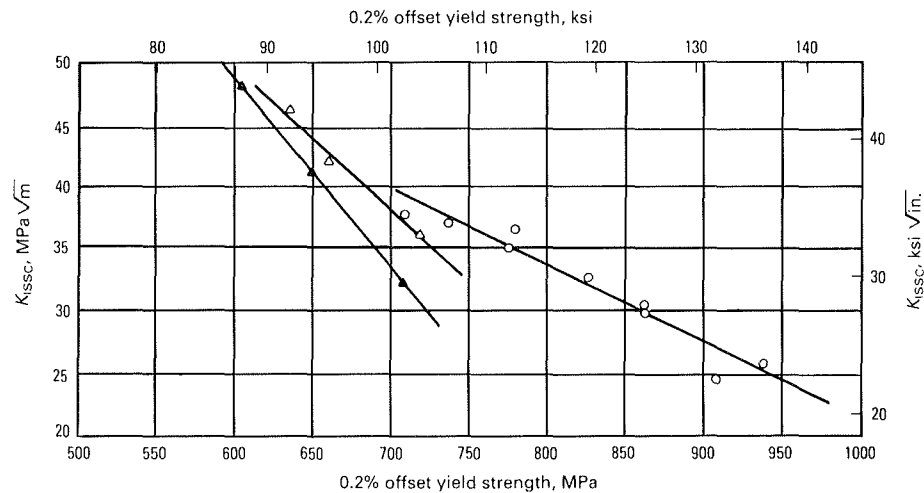
Alloy Steels

4135 Molybdenum-Niobium Modified Alloy Steel: Effect of H₂S Environment on Critical Stress

73



(a)

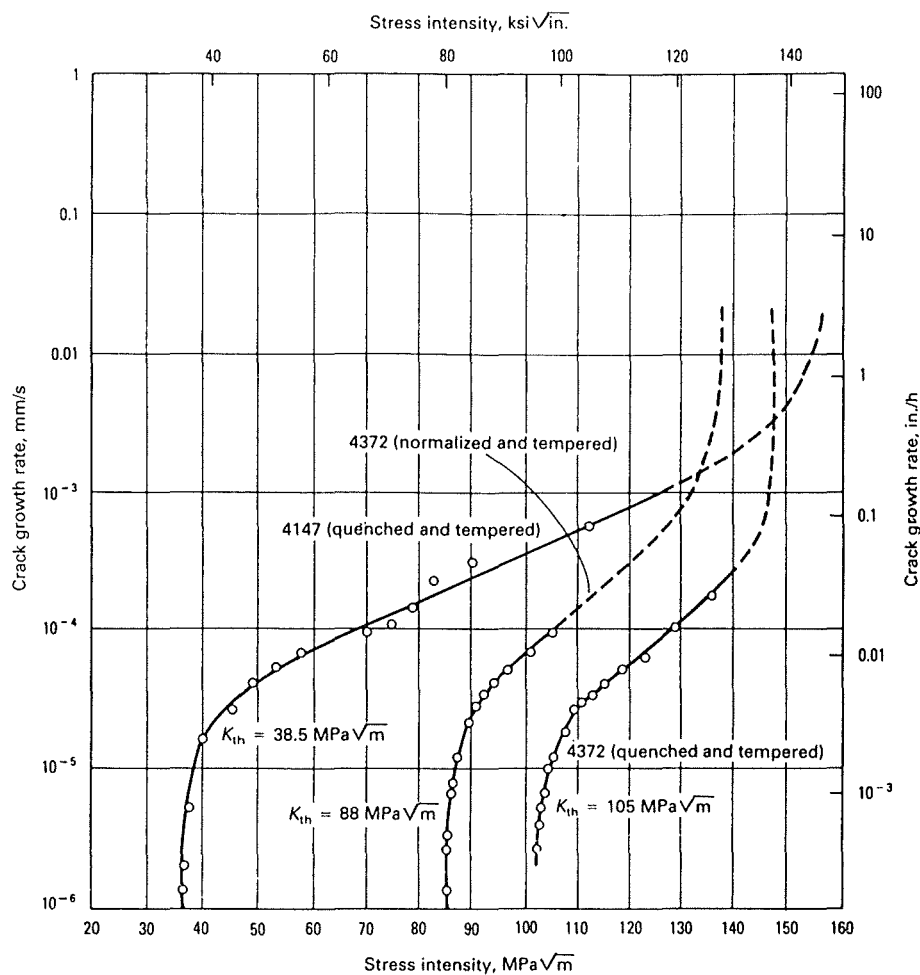


(b)

Effect of a H₂S environment on the critical stress, S_c , and K_{Isc} of molybdenum-niobium modified 4135 steel.

The data in (a) were obtained by using three-point bend specimens by determining the critical stress for cracking. The data in (b) were obtained by using wedge-loaded double cantilever beam specimens. The yield strengths were varied by cooling at different rates from the austenitizing temperature.

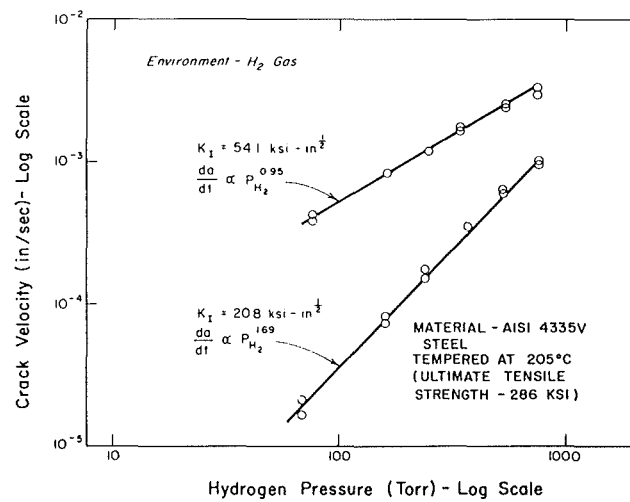
4147 and 4372 Alloy Steels: Crack Growth Kinetics in Hydrogen



Crack growth kinetics of three steels in high-pressure hydrogen at 21 MPa (3000 psi). In these tests, a plateau was not observed.

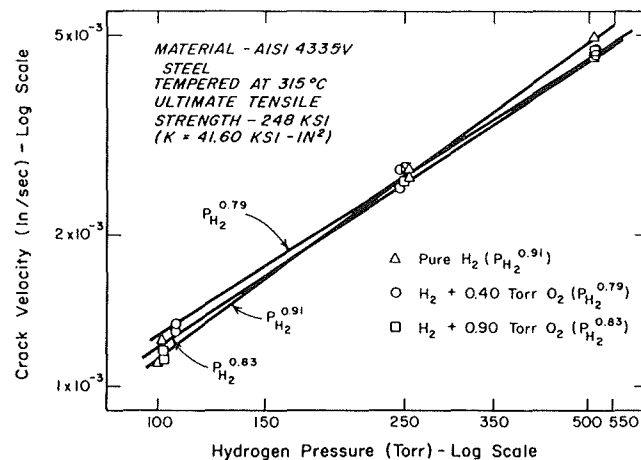
4335V Alloy Steel: Crack Velocity as a Function of Stress-Intensity, in Salt Water, H₂, and in H₂S

75



Crack velocity as a function of stress-intensity for AISI 4335V steel (tempered at 205 °C) in salt water, H₂ gas, and H₂S gas environments.

4335V Alloy Steel: Effect of Oxygen on Crack Velocity

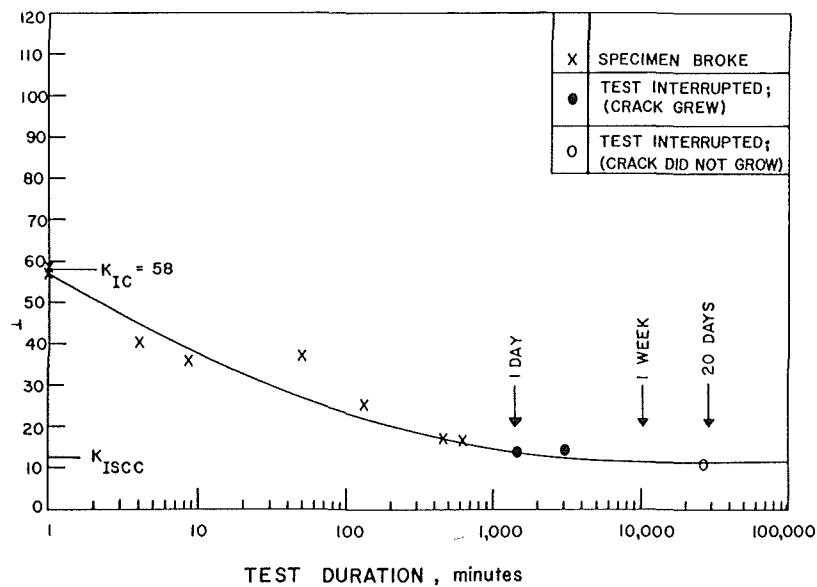


Effect of oxygen additions on the crack velocity in hydrogen gas for AISI 4335V steel tempered at 315 °C.

The absolute oxygen level is more significant than the O_2/H_2 ratio. The addition of 0.9 torr O_2 to 100 torr H_2 gas (0.9 vol%) showed no significant effect on the crack growth rate as shown. No crack growth was observed, however, with the addition of 4.7 torr O_2 to 665 torr H_2 (0.7%) for times up to 30 min.

4340 Alloy Steel: Stress-Corrosion Cracking Tests, in Sodium Chloride

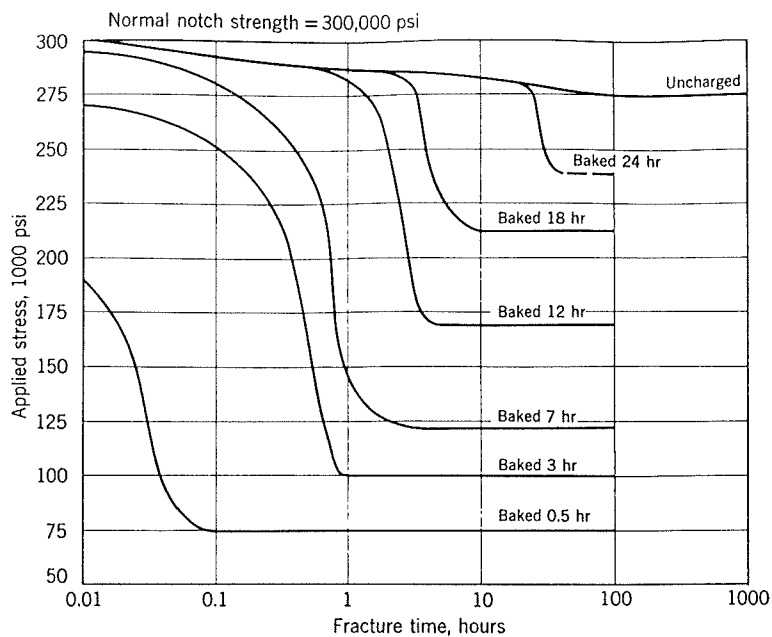
77



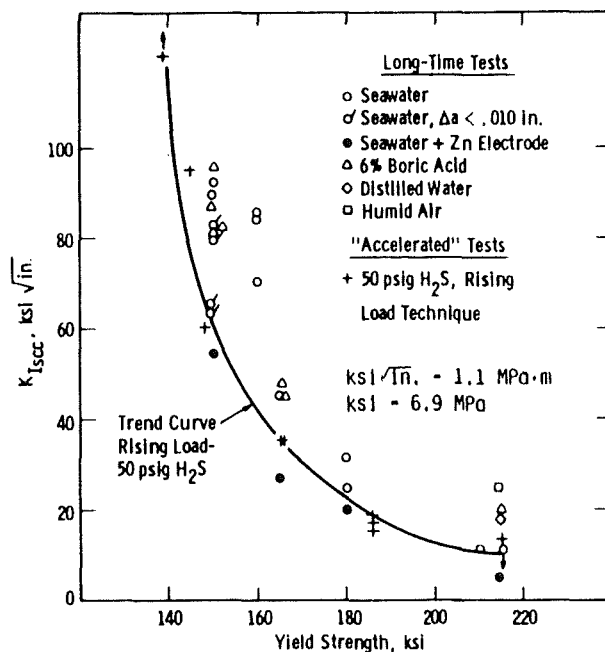
Stress-corrosion cracking tests of AISI 4340 steel, using surface cracked specimens in 3.5% NaCl solution. Similar values for K_{ISCC} were obtained with cantilever beam specimens and center-cracked specimens.

Source: C.D. Beachem and B.F. Brown, A Comparison of Three Pre-cracked Specimens for Evaluating the Susceptibility of High-Strength Steel to Stress Corrosion Cracking, in Stress Corrosion Testing, STP 425, American Society for Testing and Materials, Philadelphia, 1967, p 31-40.

4340 Alloy Steel: Fracture Time as a Function of Hydrogen Content



Delayed fracture times and minimum stress for cracking of 0.4% carbon steel as a function of hydrogen content. Specimen initially charged cathodically, baked at 150 °C for varying times to reduce hydrogen content.

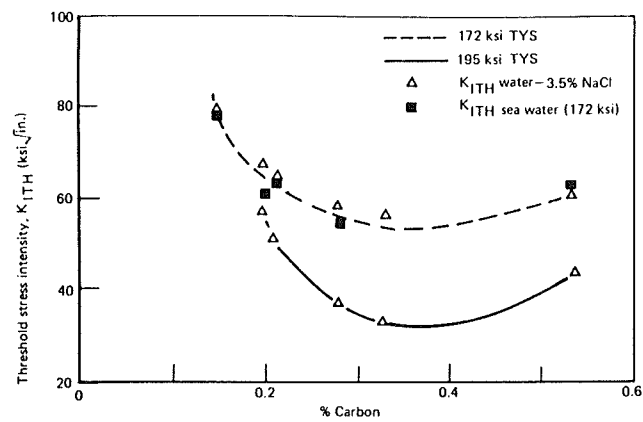


K_{Isc} values of AISI 4340 steel as determined by long-time steady load tests and by rising load tests.

These results demonstrate that the susceptibility of AISI 4340 steel increases with an increase in strength level. It was noted that corrosion fatigue crack growth rates were enhanced considerably in the H_2S environment at stress-intensity levels well below K_{Isc} so that an apparent immunity under sustained loading does not imply immunity under cyclic loading.

80

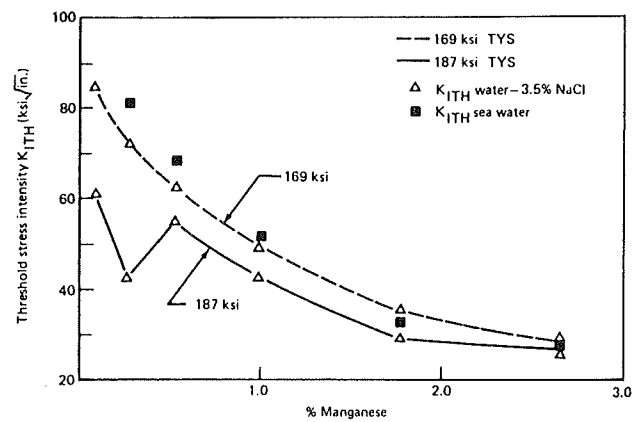
4340 Alloy Steel: Effects of Carbon Content on Resistance to Stress-Corrosion Cracking, in Water and in NaCl + Water



Effects of carbon content on resistance to stress-corrosion cracking of AISI 4340 steels quenched and tempered to either 172 or 195 ksi yield stress.

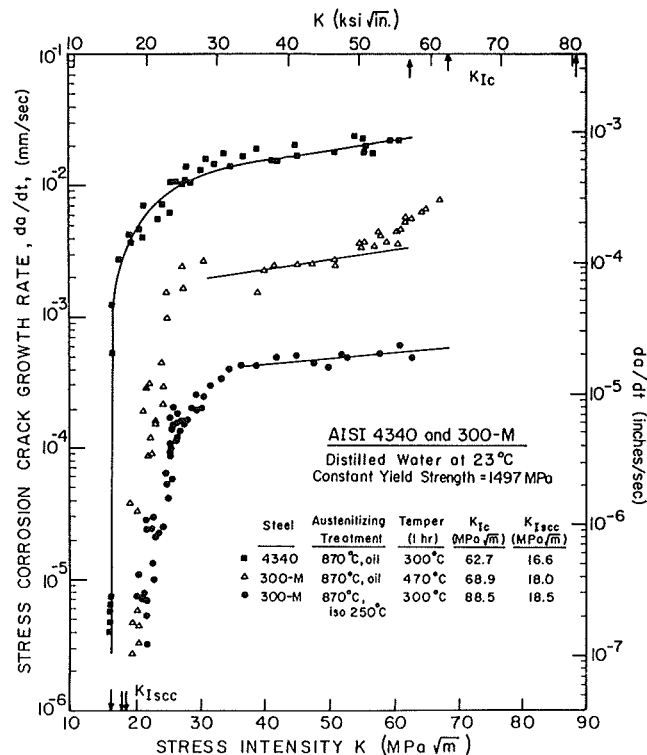
4340 Alloy Steel: Effects of Manganese Content on Stress-Corrosion Cracking Resistance, in Water and in NaCl + Water

81



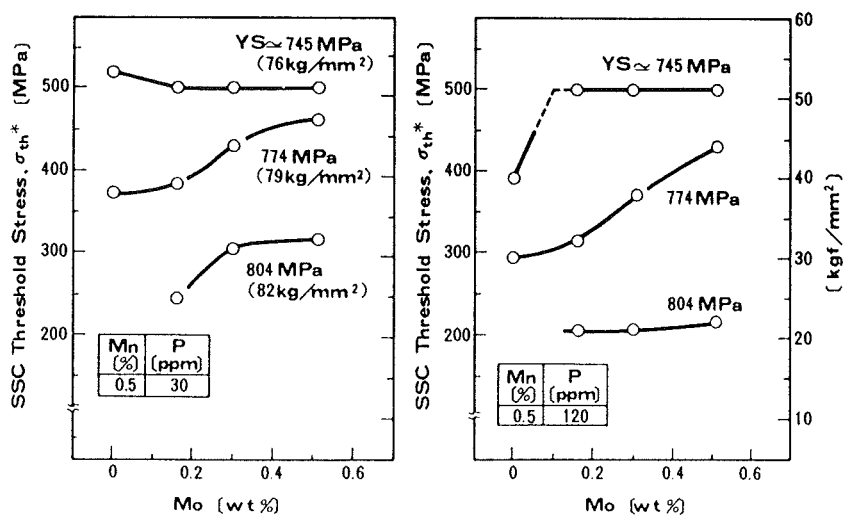
Effects of manganese content on stress-corrosion cracking resistance of AISI 4340 steels quenched and tempered to 169 and 187 ksi yield strength.

4340 and 300-M Alloy Steel: Variation of Stress-Corrosion Cracking with Stress-Intensity, in Distilled Water



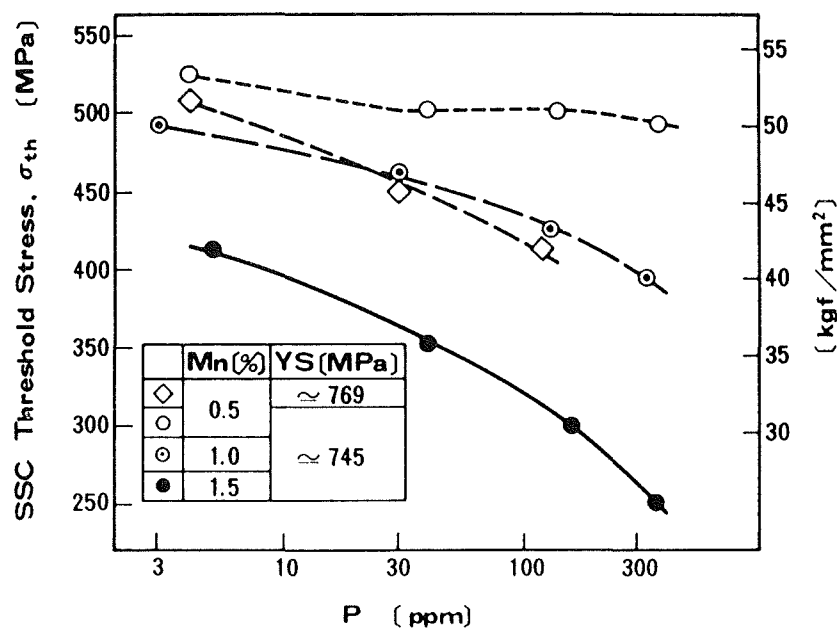
Variation of stress-corrosion crack growth rate with stress-intensity for 4340 and 300-M steel, heat treated by quenching and tempering and by isothermal transformation to produce identical yield strength. Tests were performed in distilled water at ambient temperature.

The isothermally transformed 300-M steel contains 12% retained austenite, which in this case appears to play a beneficial role in resisting hydrogen embrittlement. Quenched and tempered 300-M and 4340 contain less than 2% retained austenite.



Effect of molybdenum content on sulfide stress-corrosion cracking threshold stress of smooth steel specimens.

High-Strength Low-Alloy Steel: Effect of Phosphorus on Sulfide Stress-Corrosion Cracking

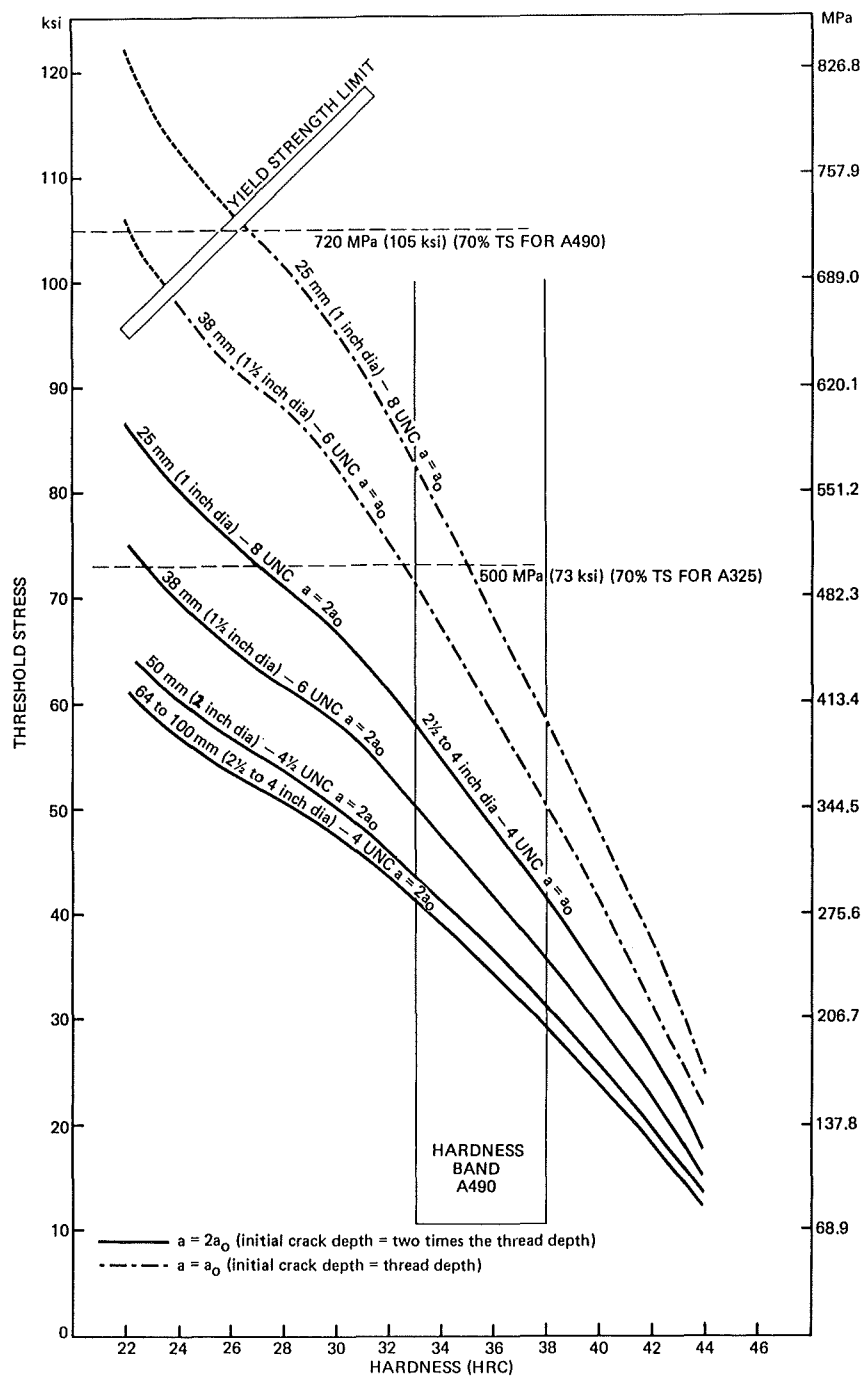


Effect of phosphorus on the sulfide stress-corrosion cracking threshold stress of smooth specimens of steel with varied manganese content.

Source: H. Asahi, et al., Effects of Mn, P, and Mo on Sulfide Stress Cracking Resistance of High Strength Low Alloy Steels, Met. Trans. A, 19A, September 1988, p 2171-2178.

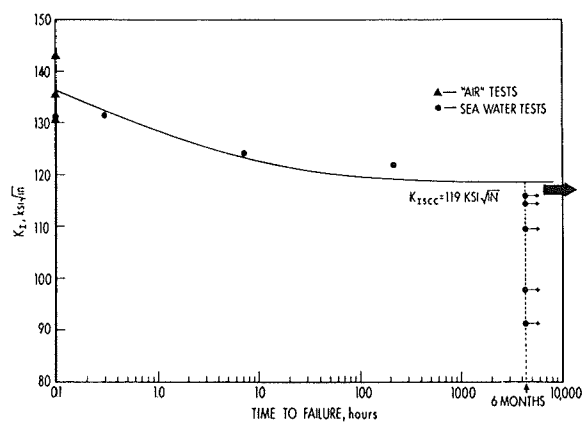
Low-Alloy Bolts: Threshold Stress for Stress-Corrosion Cracking

85

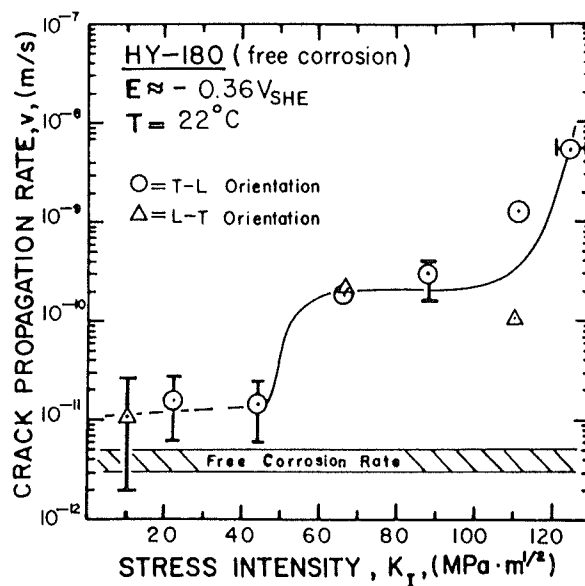


Plot of threshold stresses for stress-corrosion cracking in low-alloy quenched and tempered bolting materials with different hardness levels. Note: UNC denotes unified coarse thread.

HY-130 High-Strength Alloy Steel: Effect of Seawater on Time-to-Failure



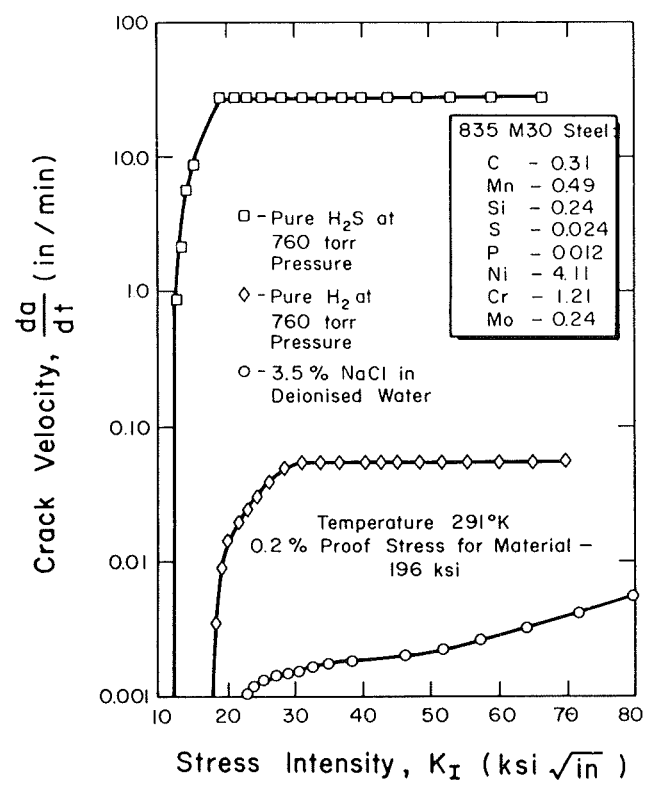
Effect of seawater environment on the time-to-failure of HY-130 steel. HY-130 is a 5Ni-Cr-Mo-V steel. The specimens were fatigue precracked cantilever beams.



Rate of stress-corrosion crack growth as a function of K_I for HY-180 steel in 3.5% NaCl solution under freely corroding conditions ($E \approx -0.36 V_{SHE}$).

The cracking velocity at the plateau was two orders of magnitude higher when the specimen was coupled to zinc ($E \approx -0.82 V_{SHE}$). Cracking was consistent with a hydrogen embrittlement mechanism, irrespective of potential.

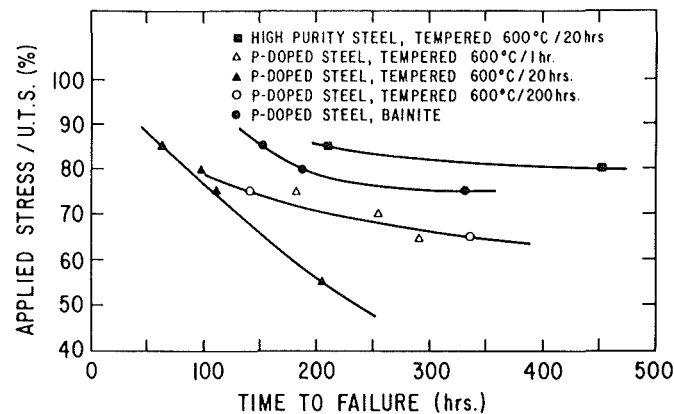
835 M30 Alloy Steel: Effect of Stress-Intensity on Crack Velocity, in NaCl, H₂, and in H₂S



Effect of stress-intensity on crack velocity for 835 M30 steel in aqueous NaCl, gaseous H₂, and gaseous H₂S environments.

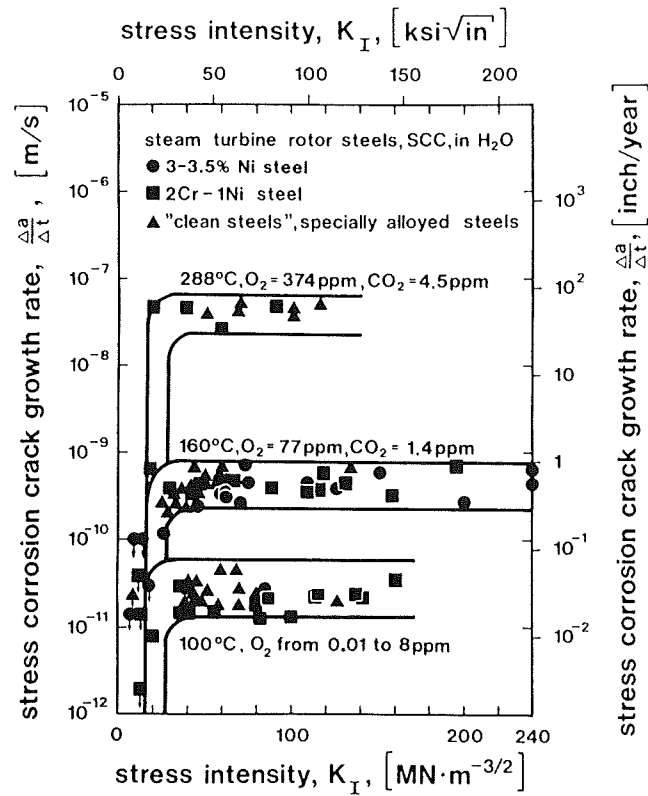
Ni-Cr-Mo-V Rotor Steels: Caustic Stress-Corrosion Cracking Behavior, in NaOH

89



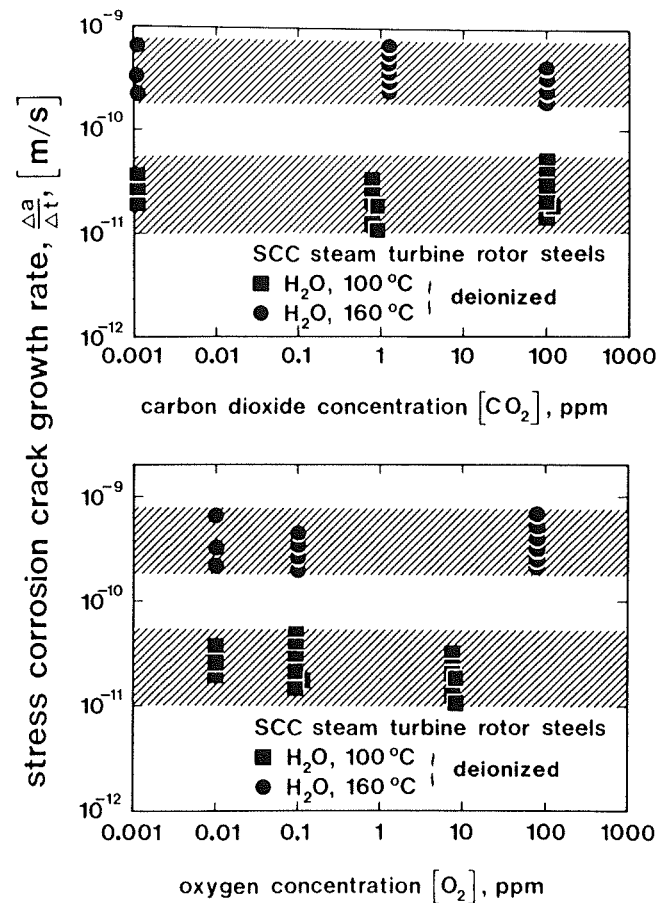
Caustic stress-corrosion cracking behavior of Ni-Cr-Mo-V rotor steels.

The specimens were tested in 9M NaOH at 98 °C at a controlled potential of -442 mV_{SHE}. The results indicate a detrimental effect due to segregation of phosphorus to grain boundaries. Heat treatments that produced large, chromium-rich carbides were beneficial. These carbides are preferentially corroded and cause pits to form at the crack tip. The pits cause crack-tip blunting and slow the rate of crack propagation.



Effect of temperature and stress-intensity on the growth rates of stress-corrosion cracks in steam turbine rotor steels.

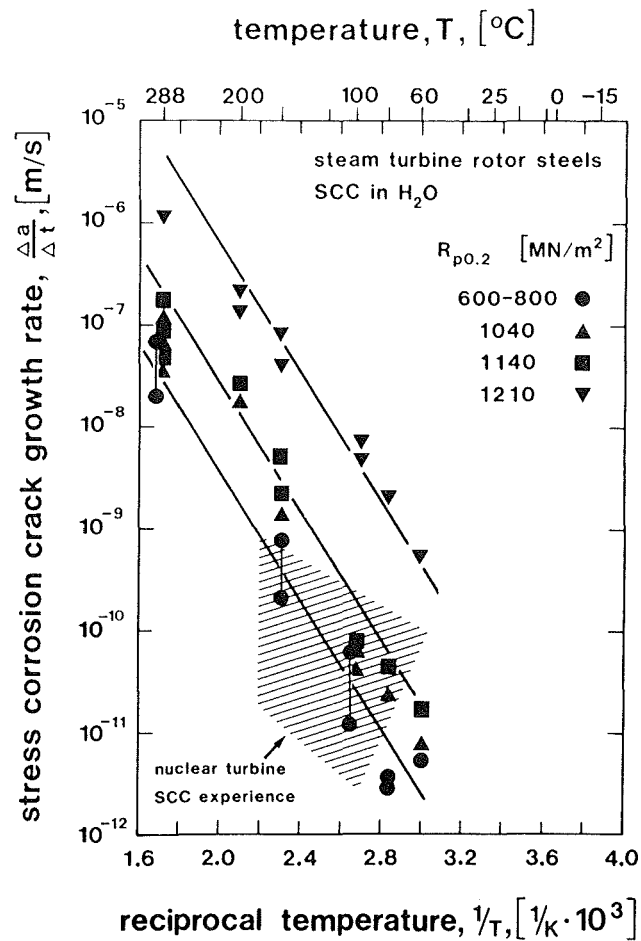
Stress-corrosion cracking in turbine rotor steels is intergranular. The plateau stress-corrosion crack growth rate is strongly affected by the yield strength of the steels and the temperature of the environment, but not by the dissolved oxygen and carbon dioxide concentrations within the range studied. Within the composition range from superclean steels to conventional steam turbine rotor steels, the elements manganese, molybdenum, nickel, phosphorus, and sulfur had no influence on the stress-corrosion crack growth rate in hot water. Fracture toughness, as modified by alloy composition and temper embrittlement, had no influence on the growth rate of stress-corrosion cracks in water.



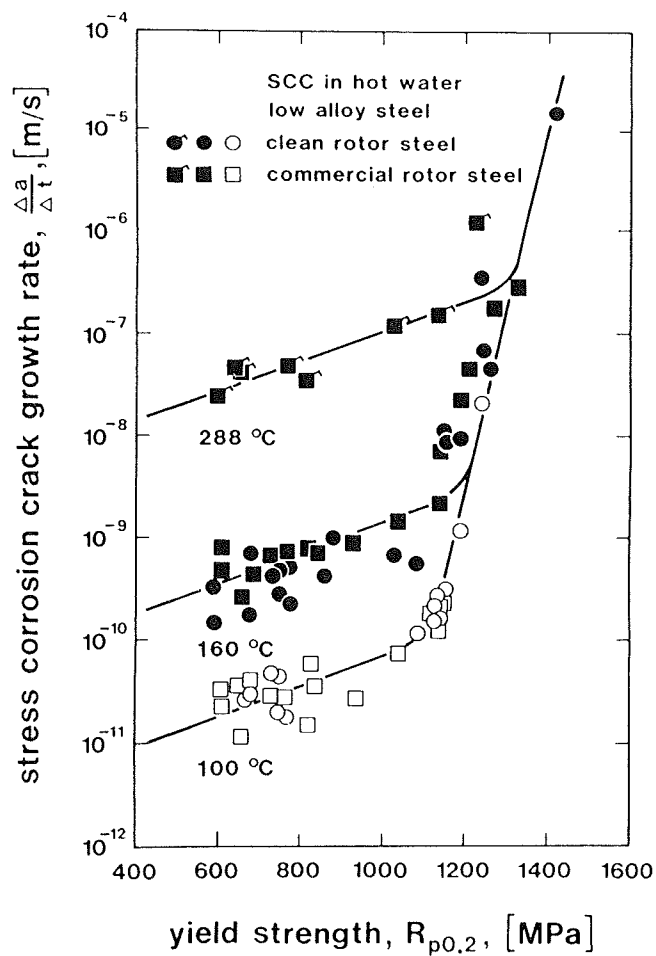
Effect of gases, dissolved in hot water, on the growth rates of stress-corrosion cracks in steam turbine rotor steels. The applied stress-intensities range between 40 and 80 $MN \cdot m^{-3/2}$ at all three temperatures.

The concentration of dissolved oxygen and carbon dioxide, within the limits investigated, does not have any influence on the stress-corrosion crack velocity.

Steam Turbine Rotor Steels: Effect of Temperature on Stress-Corrosion Crack Growth Rates in Water

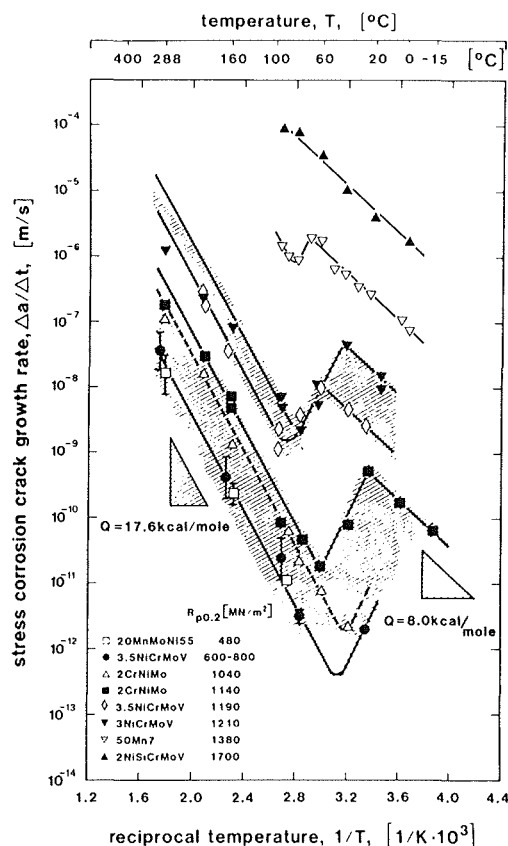


Effect of temperature on stress-corrosion crack growth rates. The shaded area indicates the service experience of steam turbine rotors. $R_{p0.2}$ is the 0.2% offset yield strength.



There are two different effects of yield strength on stress-corrosion crack growth rates: a moderate one at low strength levels and a strong effect at high strength levels. The applied stress-intensities are between 40 and 80 $\text{MN}\cdot\text{m}^{-3/2}$.

Ferritic Steels: Temperature Dependence of Stress-Corrosion Crack Growth Rates



Temperature dependence of the stress-corrosion crack growth rates observed in a large number of ferritic steels.

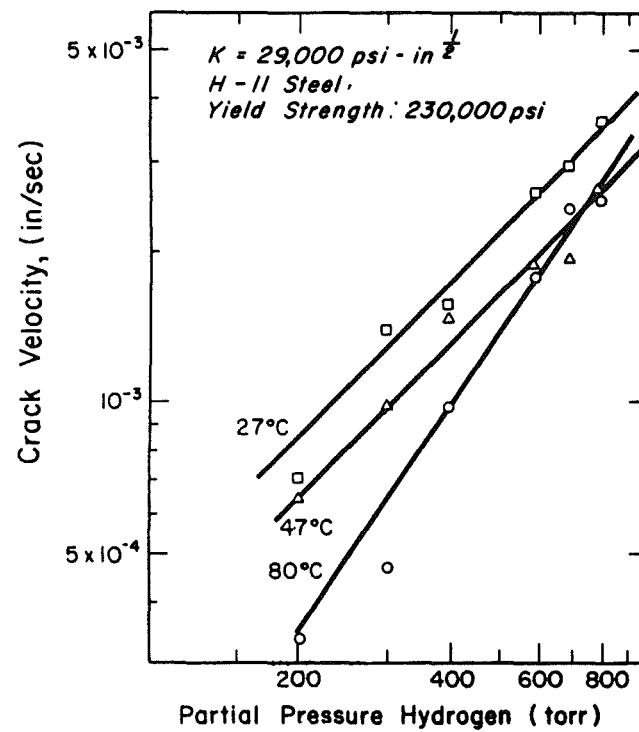
The stress-corrosion crack growth rate usually increases with increasing temperature. There are two branches of the curve where the temperature dependence characterized by the activation energy is different: one branch for high temperatures, with an activation energy of 17.6 kcal/mol, and one branch for low temperatures, with an activation energy of 8.0 kcal/mol. The region between these two parts of the curve depends on the yield strength of the steel, and in this range, the stress-corrosion crack growth rate decreases with increasing temperature.

The three different regions of temperature dependence imply three different rate-controlling processes in the stress-corrosion crack growth mechanism. The low-temperature branch, characterized by an activation energy of about 8.0 kcal/mol (~ 36 kJ/mol) has been extensively studied by Wei and co-workers and is best understood. In this temperature region, the reaction of the steel surface with water is the cause for stress-corrosion cracking. The corresponding activation energy has the same value as the activation energy for the nucleation and growth of iron oxide in water. This suggests that the hydrogen responsible for the subcritical cracking is produced during this oxidation reaction. Thus, in this temperature range, an oxidation reaction is the rate-limiting process for crack growth.

In the temperature range 60 to 288 °C, subcritical crack growth has been discussed by Ford in terms of a film rupture and anodic dissolution mechanism. Thus, it appears that there are two different thermally activated processes involved; hydrogen embrittlement at low temperatures and anodic dissolution at high temperatures. The inverse temperature dependence at intermediate temperature ranges could then be considered a transition between the two mechanisms.

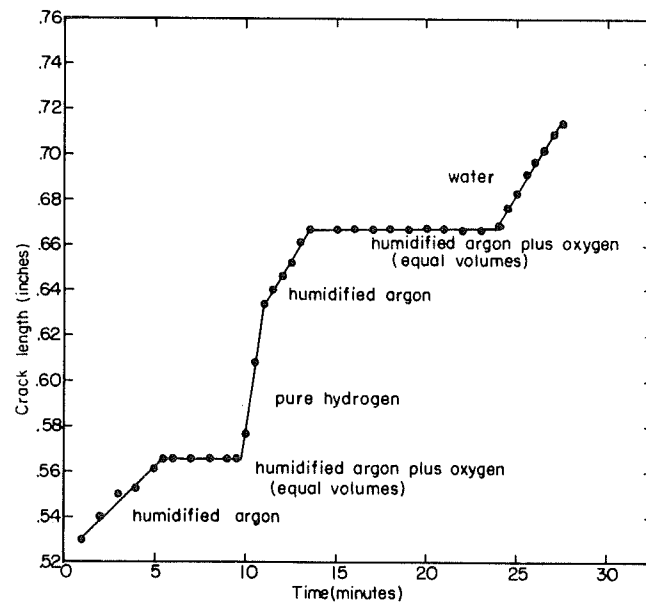
H-11 Alloy Steel: Effect of Temperature on Crack Velocity in H₂ Gas

95



Effect of test temperature on the pressure dependence of crack velocity for H-11 steel in H₂ gas.

H-11 Alloy Steel: Effect of Environment on Crack Length

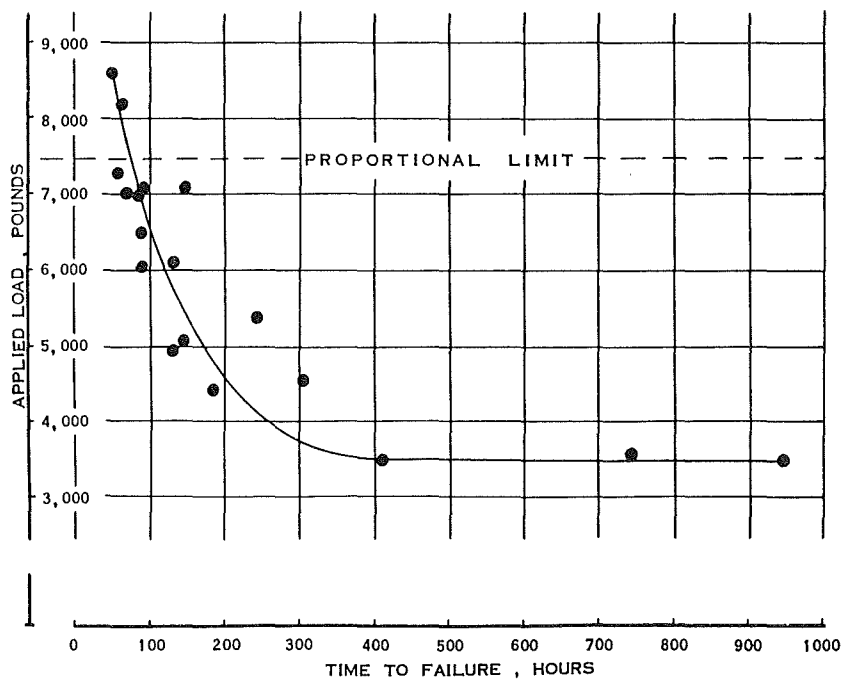


Effect of environment on crack length in a high-strength martensitic steel as a function of time.

Brittleness in this high-strength steel (H-11) can result from exposure to hydrogen at atmospheric pressure due to an adsorbed layer of atomic hydrogen at the crack tip, which provides the damaging hydrogen. Subcritical crack growth due to hydrogen gas or water vapor can be arrested if small quantities of oxygen are introduced (as little as 0.6% by volume) and a protective oxide barrier is formed. However, dissolved oxygen in water has little or no influence on the crack growth rate.

H-11 Alloy Steel: Effect of Applied Load on Stress-Corrosion Cracking Susceptibility in Sodium Chloride

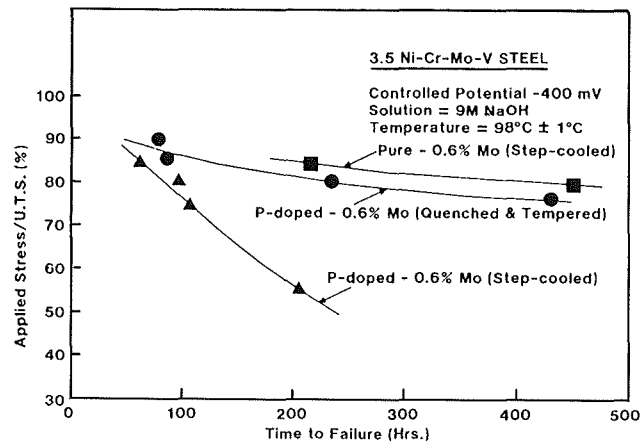
97



Effect of applied load on the stress-corrosion cracking susceptibility of $\frac{1}{4}$ -28 by 2 uncoated H-11 bolts.

H-11 is a 5.1Cr-1.3Mo-0.4C steel. The bolts had an ultimate strength of 290,500 psi with threads rolled after heat treatment. All failures occurred in the bolt shank as the result of intermittent exposure to a 3.5% NaCl solution.

Ni-Cr-Mo-V and Ni-Cr-V Phosphorous-Doped Alloy Steel: Variation of Time-to-Failure with Initial Stress, in NaOH

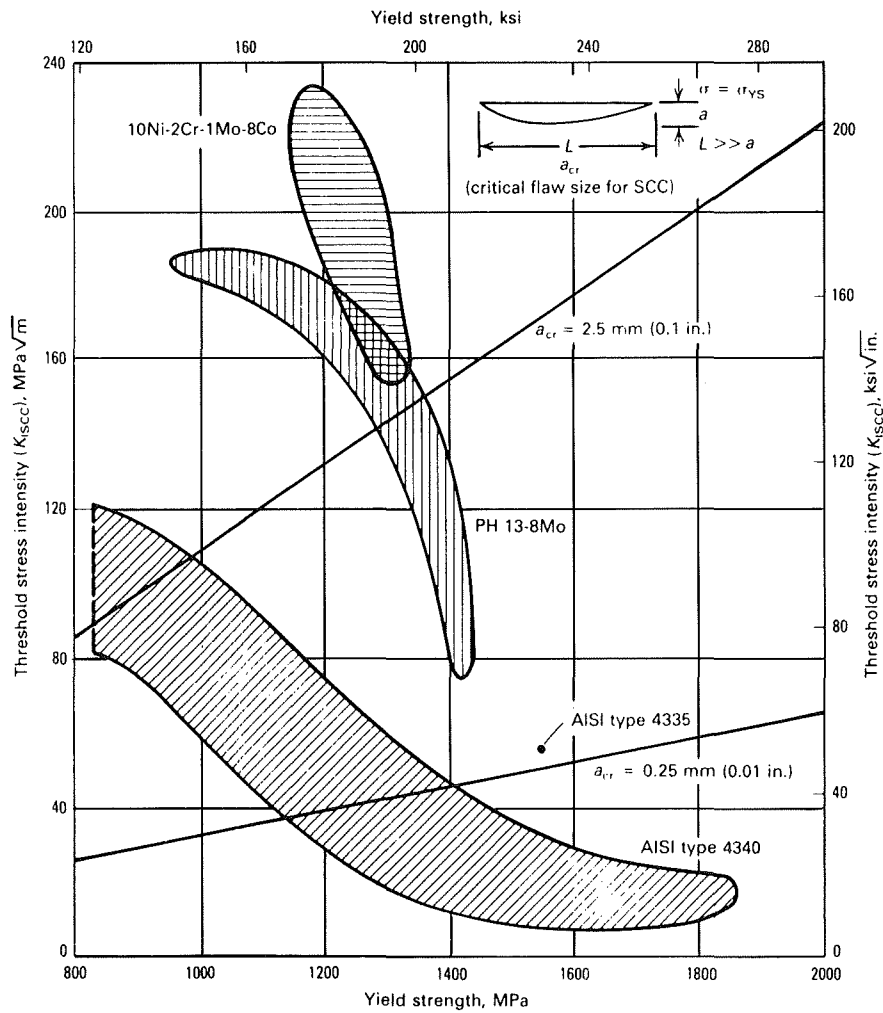


Variation of time-to-failure due to caustic stress-corrosion cracking as a function of the initial applied stress for phosphorus-doped NiCrMoV steel and a phosphorus-doped NiCrV steel.

Molybdenum and phosphorus, probably as anions precipitated from solution, aid in passivating the sides of the crack and thus help to keep the crack tip sharp. This sharpness will increase the speed with which the crack will propagate. Removal of the molybdenum greatly increases the number of cracks that nucleate. This higher crack density would increase the relative area of the anode to the cathode and thus act to decrease the crack growth rate.

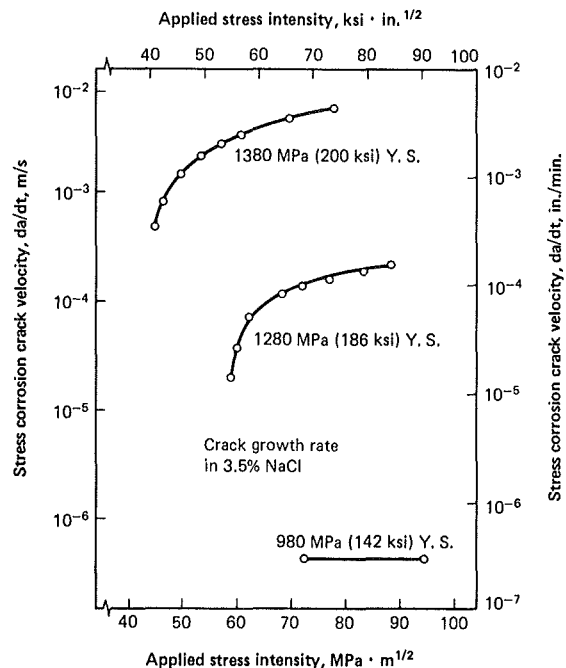
High-Strength Alloy Steels: Comparison of Stress-Corrosion Cracking Behavior of 10Ni-2Cr-1Mo-8Co, PH 13-8Mo, 4335, and 4340, in Salt Water

99

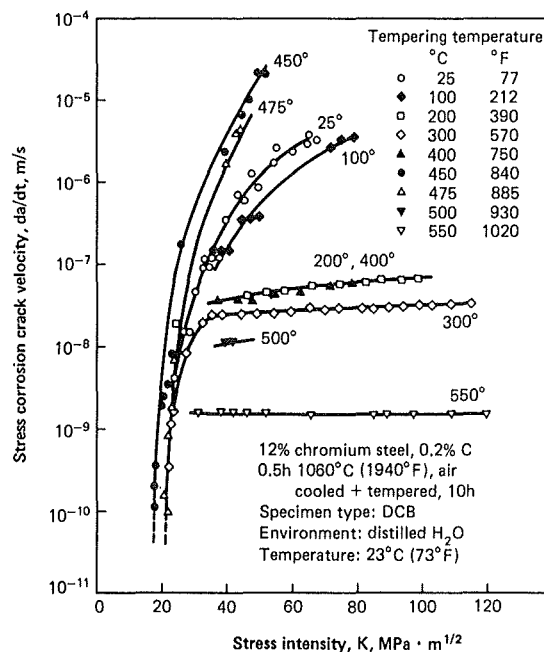


A comparison of stress-corrosion cracking behavior of several high-strength steels based on the threshold stress-intensity K_{Isc} values in salt water.

4340 Alloy Steel and 12% Chromium Steel: Effect of Tempering Temperature on Stress-Corrosion Cracking, in Sodium Chloride and in Distilled Water



(a) Crack velocity for 4340 steel in 3.5% sodium chloride.



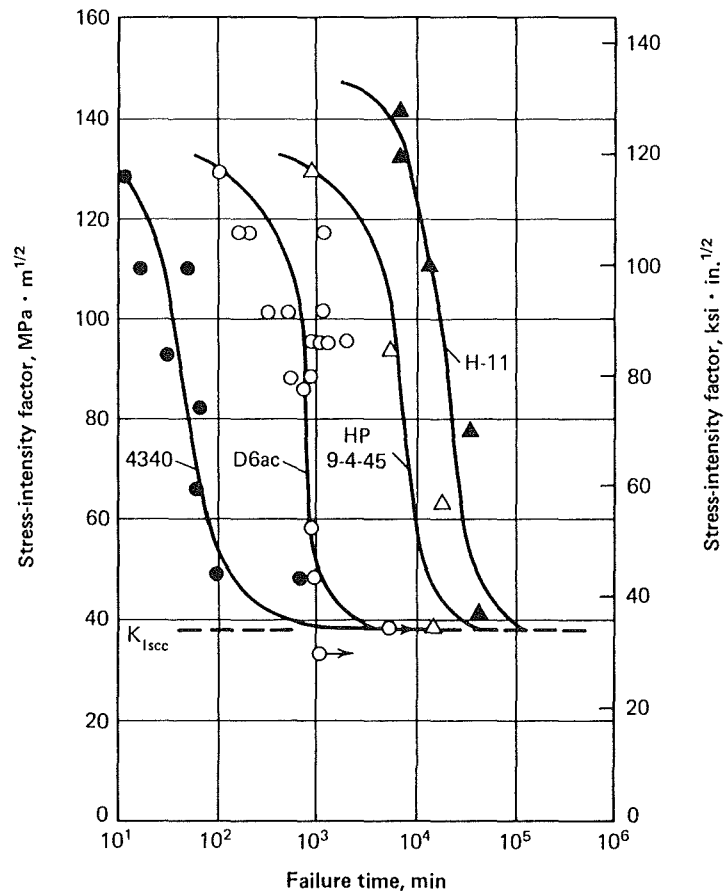
(b) Crack velocity for a 12% chromium steel in distilled water.

Effects of tempering temperature and applied stress-intensity factor on velocity of stress-corrosion cracking.

Source: R.H. Van Stone and A.W. Gunderson, Fracture Properties of Carbon and Alloy Steels, in Application of Fracture Mechanics for Selection of Metallic Structural Materials, J.E. Campbell, W.W. Gerberich, and J.H. Underwood, Ed., American Society for Metals, Metals Park, 1982, p 41-104.

High-Strength Alloy Steels: Stress-Corrosion Failure Times of 4340, D6ac, HP 9-4-45 and H-11, in Distilled Water

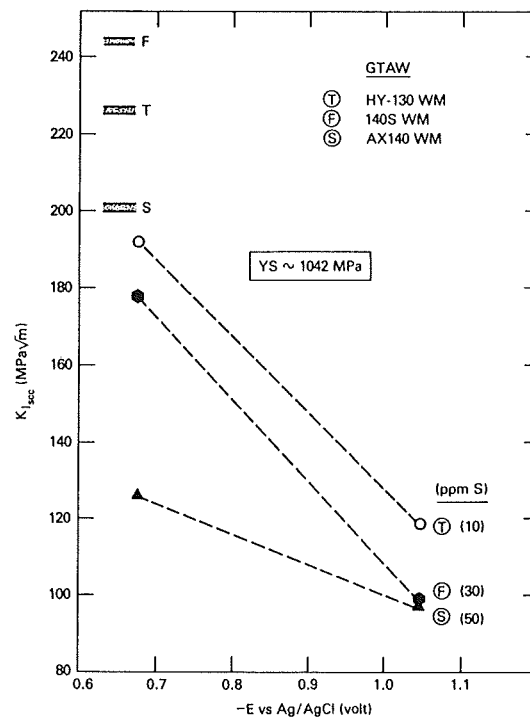
101



Comparison of the stress-corrosion failure times of four high-strength steels. The steels were subjected to sustained loading in distilled water. Each steel had been hardened and tempered to a tensile strength of about 1650 MPa (240 ksi).

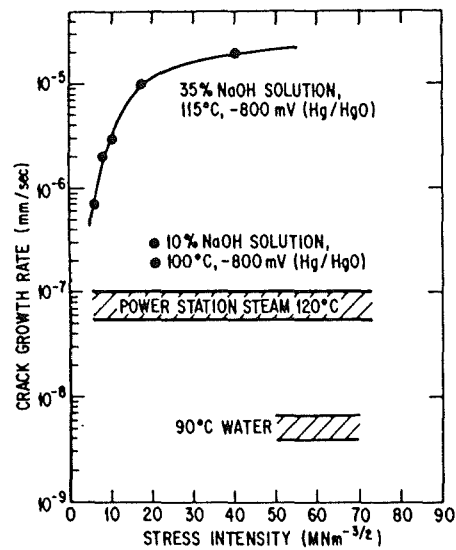
Source: R.H. Van Stone and A.W. Gunderson, Fracture Properties of Carbon and Alloy Steels, in Application of Fracture Mechanics for Selection of Metallic Structural Materials, J.E. Campbell, W.W. Gerberich, and J.H. Underwood, Ed., American Society for Metals, Metals Park, 1982, p 41-104.

Gas Tungsten-Arc Welded Alloy Steel: Influence of Sulfur Content on Electrochemical Potential



Influence of sulfur content on the effects of electrochemical potential on K_{Isc} of gas tungsten-arc steel weld metals at a yield strength of 1042 MPa.

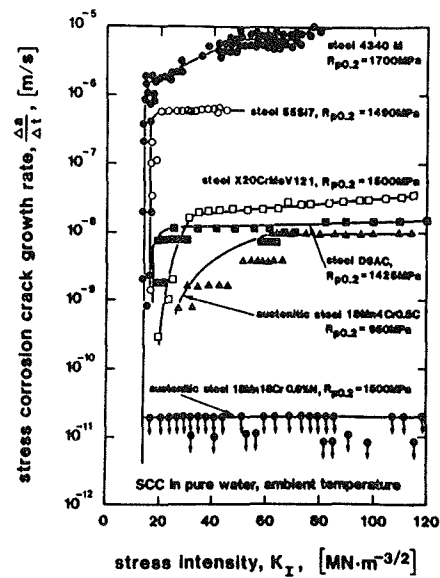
A low sulfur content appears to be a major prerequisite for improvement in the stress-corrosion cracking properties in gas tungsten-arc steel weld metals.



Average rate of stress-corrosion cracking of 3Cr-Mo steels in NaOH solutions, power station steam, and water plotted as a function of stress-intensity.

Extremely long initiation times on the order of 20,000 h have been reported in steam and water. Cracks tend to initiate in pits and then propagate along grain boundaries.

High-Strength Steel: Comparison of Stress-Corrosion Cracking Resistance in Water

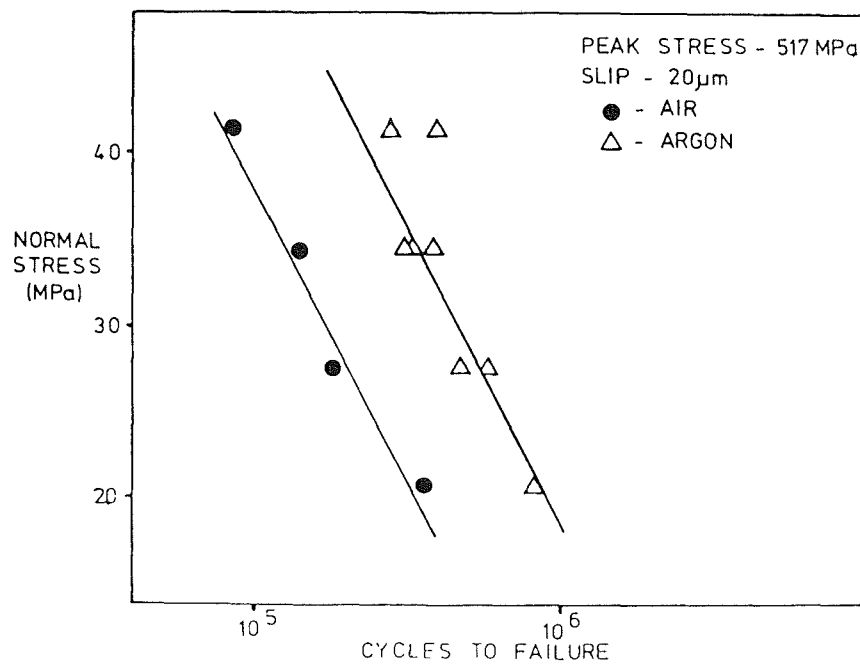


Comparison of the stress-corrosion cracking resistance of several high-strength steels in water at ambient temperature.

All of the high-strength steels are susceptible to stress-corrosion cracking in water at ambient temperature, with the exception of a high-nitrogen, low-carbon austenitic steel.

4130 Alloy Steel: Effect of Atmosphere on Fretting Fatigue Life

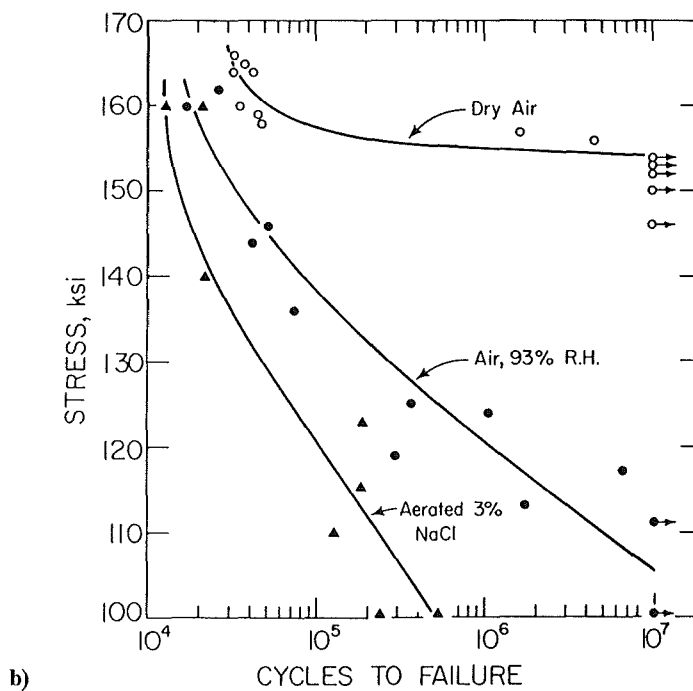
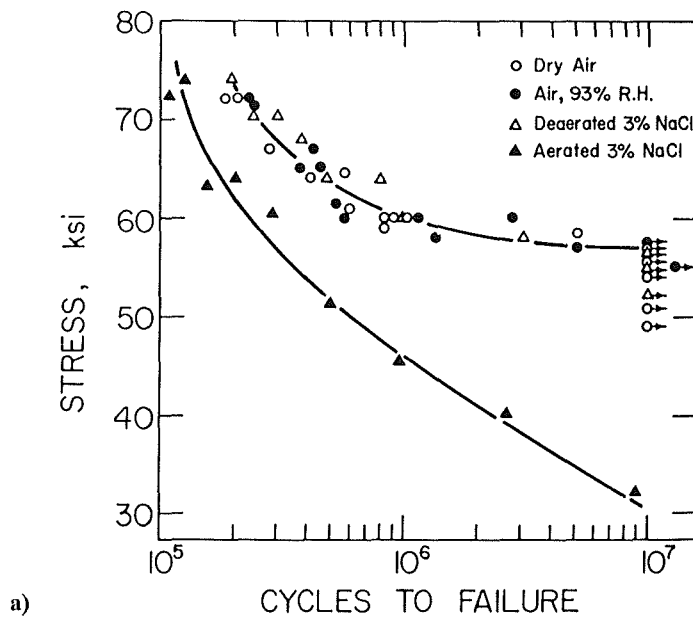
105



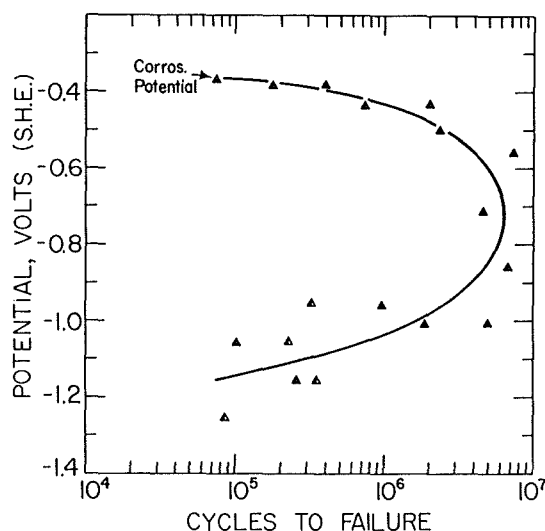
Effect of atmosphere on fretting fatigue life.

Fatigue tests were performed at $R \approx 0$. The normal stress refers to the pressure of the fretting pad on the surface of the specimen.

4140 Alloy Steel: Effect of Moisture and Dissolved Oxygen on Reversed Bending Fatigue Life, in Sodium Chloride



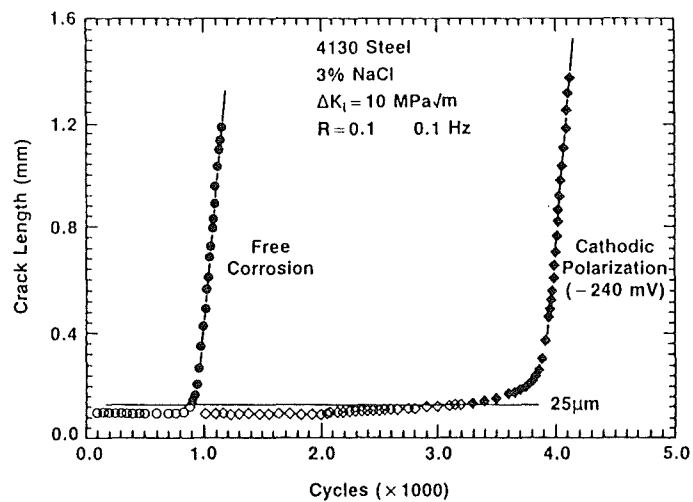
Reversed bending fatigue life of 4140 steel as influenced by hardness level and environment.



Effect of applied potential on the reversed bending fatigue life of 4140 steel at a hardness level of 52 HRC stressed at 141 ksi (i.e., below the fatigue limit) in dry air.

In aerated 3% NaCl solution, as the potential was made more active (greater cathodic current) cathodic protection was provided and a maximum increase in fatigue life was obtained at $-0.5 V_{SHE}$. At still more active potentials, the fatigue life decreased, and at $-1.25 V$ hydrogen was visibly evolved at the specimen surface. Hydrogen evolution had no effect on the fatigue life at a lower hardness level of 37 HRC. An applied anodic current resulted in a decrease in life up to an anodic current density of about $2.5 \mu A/cm$. The results can be interpreted in terms of a hydrogen embrittlement mechanism. At the corrosion potential, anodic dissolution and the dissociation of water to form H^+ and OH^- ions occur, with the H^+ ions being adsorbed at local cathodic regions at the cracking sites. This hydrogen promotes the cracking process. As the potential is made more active, the rate of anodic dissolution decreases and the life increases. However, as in the case of the 4140 steel heat treated to 52 HRC, if the potential is made too active, the conditions for hydrogen evolution will be met, and the adsorption of some of this hydrogen leads to a reduction in fatigue life in a susceptible microstructure. If the potential is made less active, initially there will be a decrease in fatigue life due to an increased anodic dissolution rate and an increased rate of H^+ creation.

4130 Alloy Steel: Surface Crack Depth versus Load Cycles, in Sodium Chloride

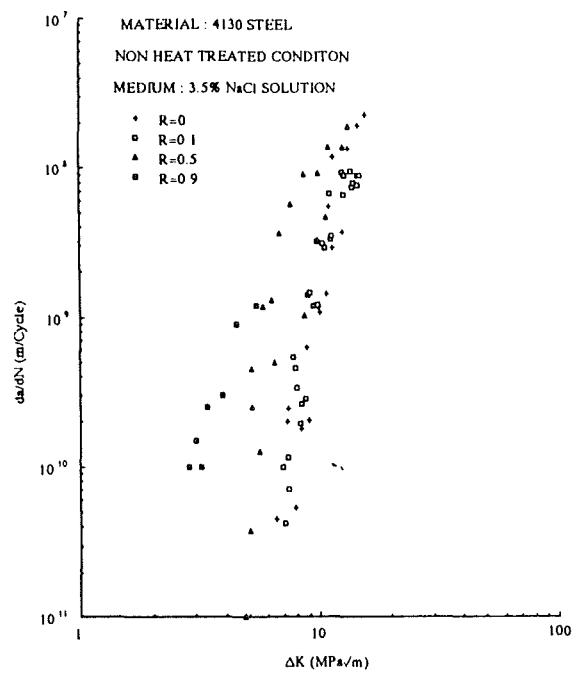


Surface crack depth versus load cycles for 4130 steel, freely corroded ($-560 \text{ mV}_{\text{SCE}}$) or cathodically polarized ($-800 \text{ mV}_{\text{SCE}}$) in aerated 3% NaCl.

Note that the primary benefit of cathodic polarization occurs when the crack is less than $40 \mu\text{m}$ in length. At greater lengths, the growth rates are approximately equal.

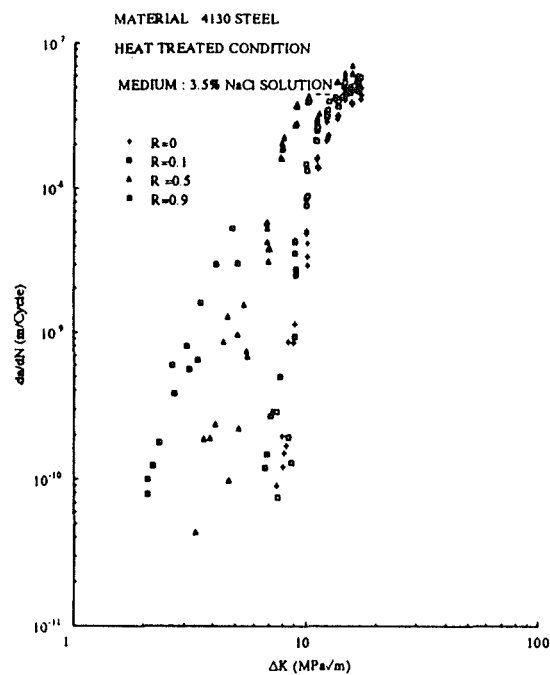
4130 Steel: Effect of Load Ratio on Corrosion Fatigue Crack Growth Rate

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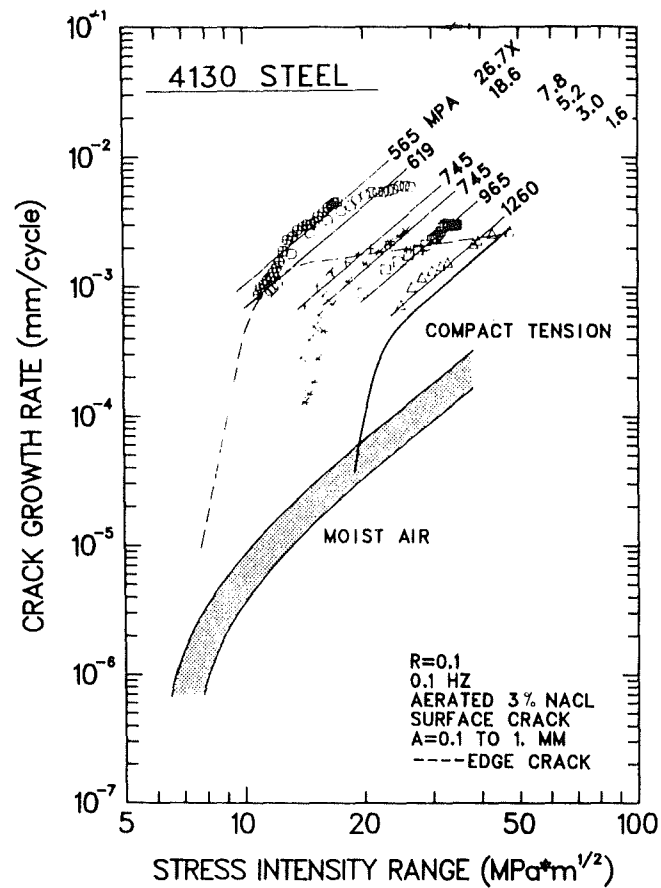
Effect of load ratio on corrosion fatigue crack growth rate in annealed 4130 steel ($\sigma = 450$ MPa).

4130 Steel: Effect of Load Ratio on the Corrosion Fatigue Crack Growth Rate



Effect of load ratio on the corrosion fatigue crack growth rate in 4130 steel quenched and tempered at 500 °C ($\sigma_y = 1330$ MPa).

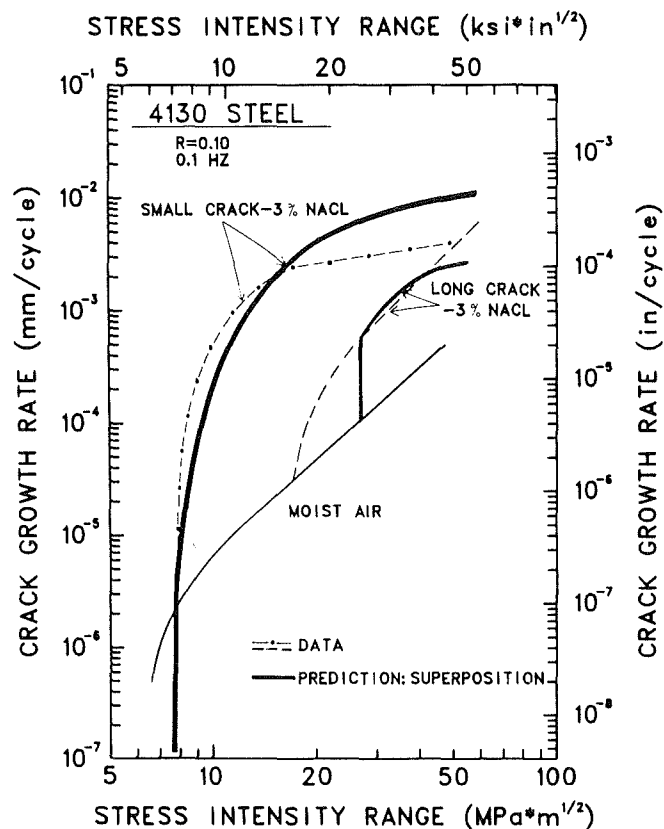
The increase in fatigue crack growth rate with increase in R was greater in the quenched and tempered as compared to the annealed condition, a reflection of the lower toughness of the quenched and tempered condition.



Effect of applied stress range on the growth of small surface cracks in salt water. The applied stress range and growth rate enhancement factors are listed.

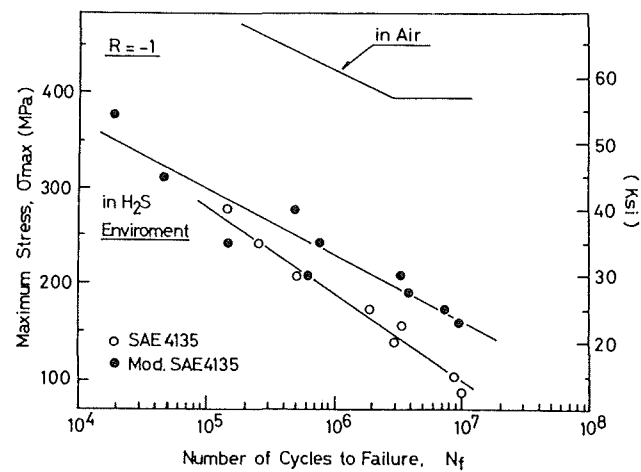
The rapid growth of small cracks was attributed to enhanced cathodic hydrogen production within the occluded crack. Increased H^+ and decreased oxygen inhibition to consume H^+ within small cracks are probable contributors to the chemical driving force. Slower crack growth rates with an increase in stress in the small crack regime correspond to an increase in the fraction of transgranular fracture, which involves a slower mode of growth, with increasing stress. Acid conditions are developed more for a short crack than a long crack due to enhanced elimination of Fe^{++} and the otherwise retarding influence of this species on the dissolution and hydrolysis reaction rates.

4130 Alloy Steel: Corrosion Fatigue Crack Growth Rates, in Sodium Chloride



Corrosion fatigue crack growth rates predicted on linear superposition of stress-corrosion cracking kinetics for short and long cracks exposed to 3% NaCl.

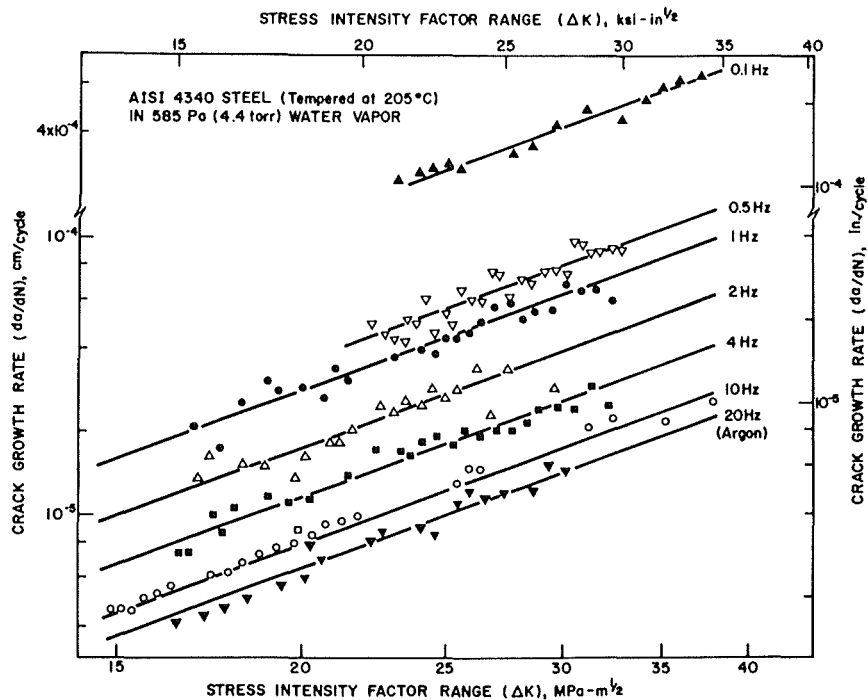
The weak ΔK dependence of the growth rate in 3% NaCl is interpreted as a characteristic of the superposition of rate-limited stress-corrosion and cycle-dependent fatigue crack growth rates. The predictions shown are based on the Wei-Landes linear superposition model for sinusoidal loading at 0.1 Hz. For long cracks, $da/dt = 4 \times 10 \text{ mm/s}$. Short crack predictions are based on static load growth data for 0.1 to 2.0 mm deep cracks. The agreement between measured and predicted rates is excellent, supporting the enhanced acidification explanation for the chemical crack size effect.



Effect of a saturated H₂S environment on the corrosion fatigue behavior of SAE 4135 steel and a modified SAE 4135 steel.

The modification consisted of increasing the molybdenum content from 0.2 to 0.75% and adding 0.036% Nb. Crack propagation behavior in H₂ was not affected by the modification, and it was concluded that the modification improved resistance to crack initiation.

4340 Alloy Steel: Room-Temperature Fatigue Crack Growth Rates, in Argon and in Water Vapor



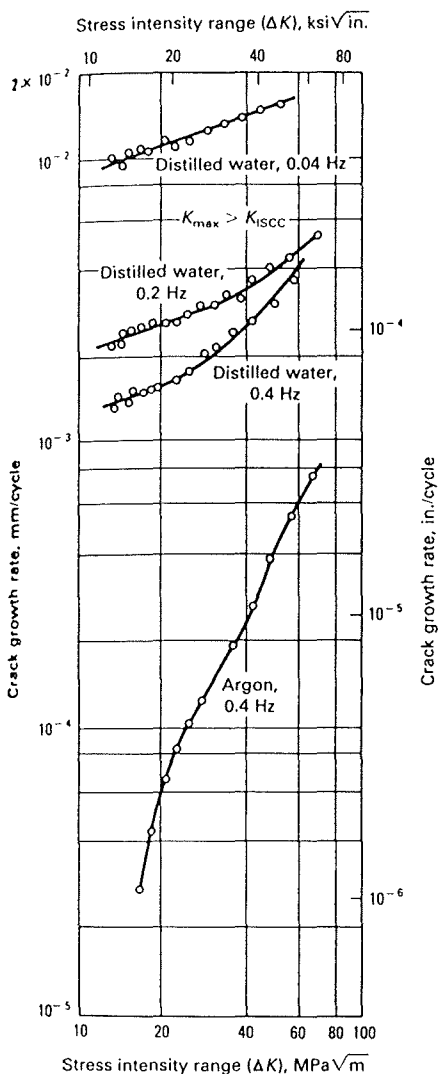
Room-temperature fatigue crack growth rates in AISI 4340 steel tested in dehumidified argon and in water vapor (below K_{Isc}) at $R = 0.1$

Wei proposed that the total rate of fatigue crack growth in an aggressive environment is made up of a component due to pure fatigue, a component due to an environmental contribution, and above K_{Isc} , a component due to stress-corrosion cracking. These contributions are linearly combined as:

$$(da/dN)_{\text{total}} = (da/dN)_{\text{fatigue}} + (da/dN)_{\text{corr}} + (da/dN)_{\text{scc}}$$

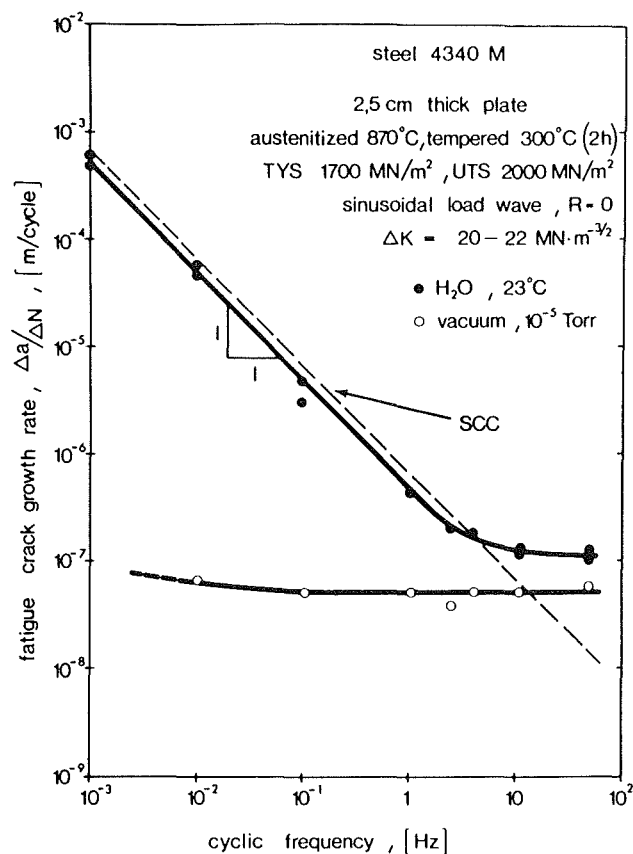
4340 Alloy Steel: Effect of Stress-Intensity Range and Load Frequency on Fatigue Crack Growth, in Distilled Water

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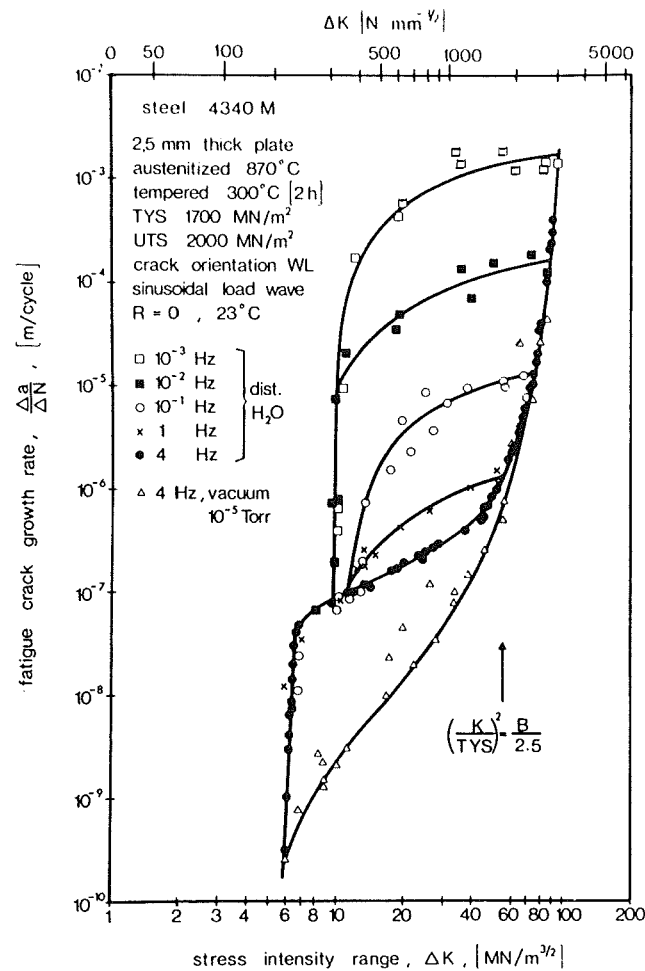
Effect of stress-intensity range and loading frequency on fatigue crack growth in ultrahigh strength 4340 steel exposed to distilled water at 23 °C.

4340 M Alloy Steel: Effect of Frequency on Growth Rate of Fatigue Cracks, in Water and in Vacuum



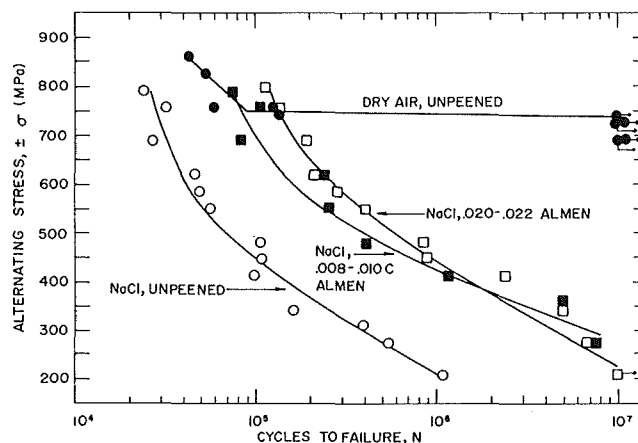
Effect of frequency on the growth rate of fatigue cracks in high-strength 4340 M steel exposed to water and vacuum.

The dashed line has been computed on the basis of a superposition of stress-corrosion crack growth rates on the frequency-independent vacuum fatigue crack growth rate. The calculated values are in good agreement with experimental results for frequencies less than 3 Hz.



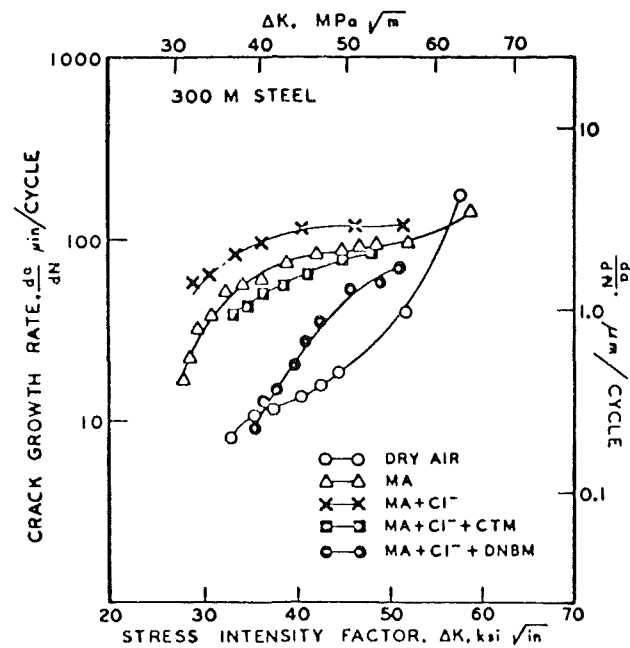
Effect of cyclic stress-intensity range and frequency on the growth rate of fatigue cracks in high-strength 4340 M steel exposed to water and vacuum.

6150 Alloy Steel: Effect of Shot Peening on Reverse Bending Fatigue Life, in Sodium Chloride



Effect of shot peening on the reverse bending fatigue life of AISI 6150 steel in 3% NaCl solution.

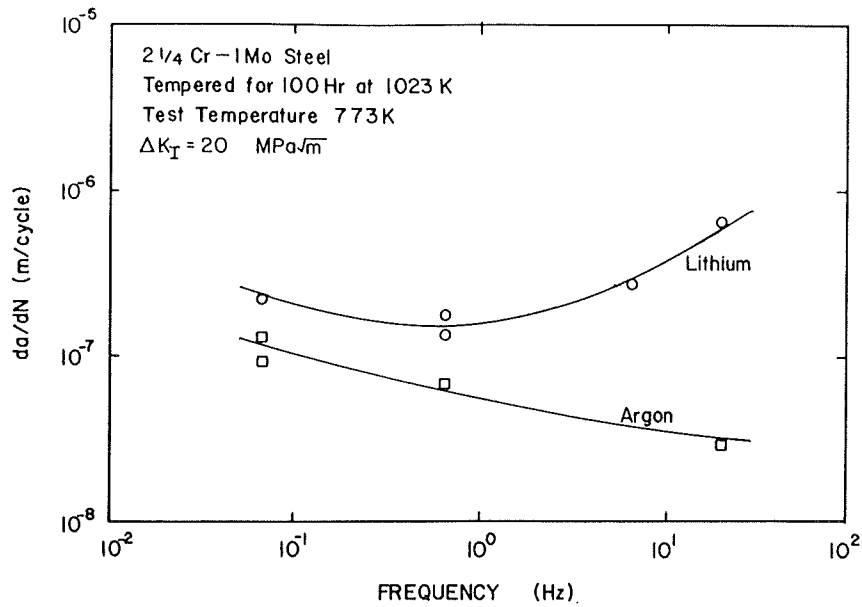
An Almen intensity of 0.020 to 0.022 A results in an estimated maximum compressive residual stress of 793 MPa, with the estimated depth of the compressive layer being 0.35 to 0.41 mm. An Almen intensity of 0.008 to 0.010 C results in a similar compressive stress, but with the depth of the compressive layer increased to 0.51 to 0.61 mm. The beneficial effect of shot peening was attributed to a significant reduction in the early fatigue crack propagation rates due to the compressive residual stress layer at the surface.



Effect of inhibitors on the crack growth rate in 300M steels.

MA = moist air; CTM = dichromate + tellurate + molybdate;
DNBM = dichromate + nitrite + borate + molybdate.

2.25Cr-1Mo Alloy Steel: Fatigue Crack Growth Rate for Argon and Lithium Environments

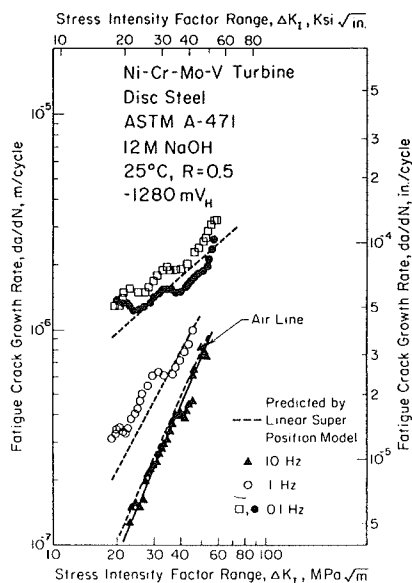


Fatigue crack growth rate as a function of frequency for argon and lithium environments at 773 K.

The variation in crack growth rate in lithium was associated with a change from intergranular fracture at low frequencies to transgranular fracture at high frequencies. Fatigue fracture in argon was transgranular, independent of frequency.

Ni-Cr-Mo-V Turbine Disc Steel: Effect of Frequency on Corrosion Fatigue Crack Growth Rate, in NaOH

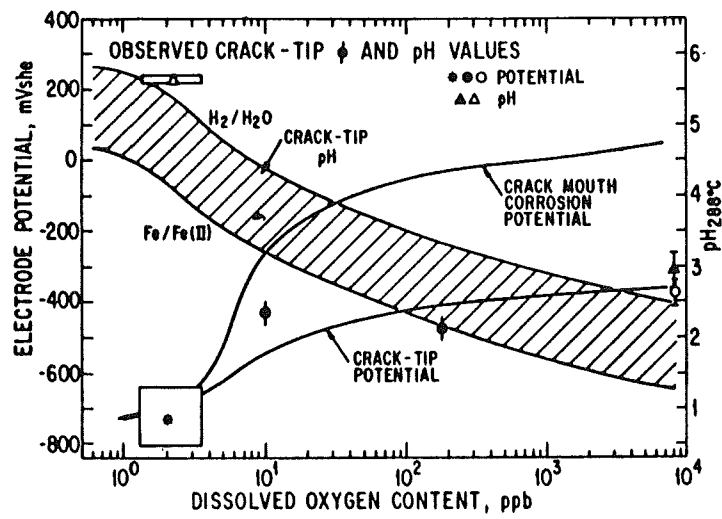
121



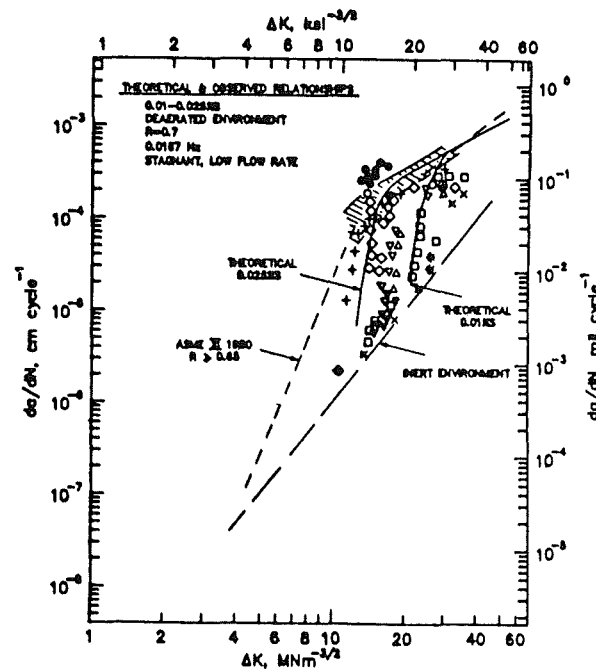
Effect of frequency on corrosion fatigue crack growth rate of ASTM A-471 turbine disc steel in room temperature 12M NaOH at $R = 0.5$ and a cathodic potential of -1280 mV_H.

Intergranular cracking was observed at cathodic potentials compared to transgranular cracking at open circuit potentials and in air. It was suggested that hydrogen diffusion into the material ahead of the crack tip was a necessary condition to accelerate cracking. Positive and negative sawtooth waveform results show that environmental effects are accelerated during the loading portion of the cycle.

Low-Alloy Steel: Variation of Crack Tip Potential, as a Function of Dissolved Oxygen Content, in Water



Variation of crack tip potential and pH for stainless steels and low-alloy steels in 288 °C water, as a function of the dissolved oxygen content in the bulk solution.

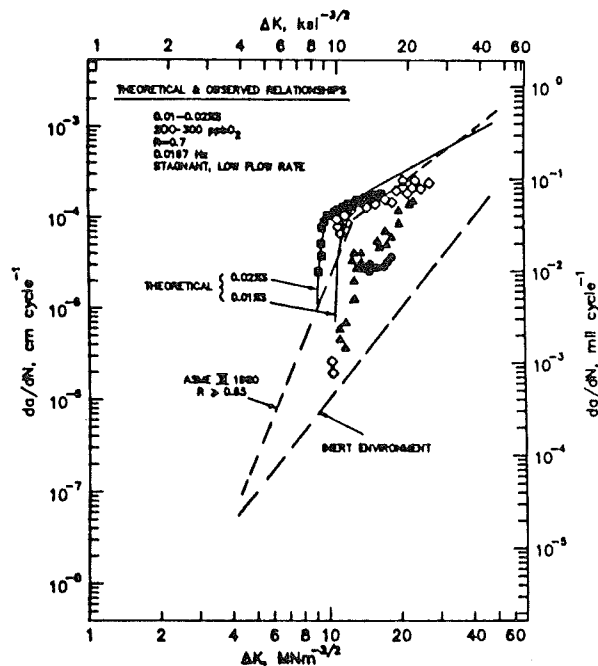


Comparison between theoretical and observed da/dN versus ΔK relationships for A533B and A508 steels with 0.01 to 0.025% S in deaerated water at 0.0167 Hz and $R = 0.7$.

The equation used in making predictions was:

$$(da/dN)_{\text{total}} = (\frac{1}{2}\text{freq})[f(n)(\epsilon_{\text{cf}})^n] + (da/dN)_{\text{inert}}$$

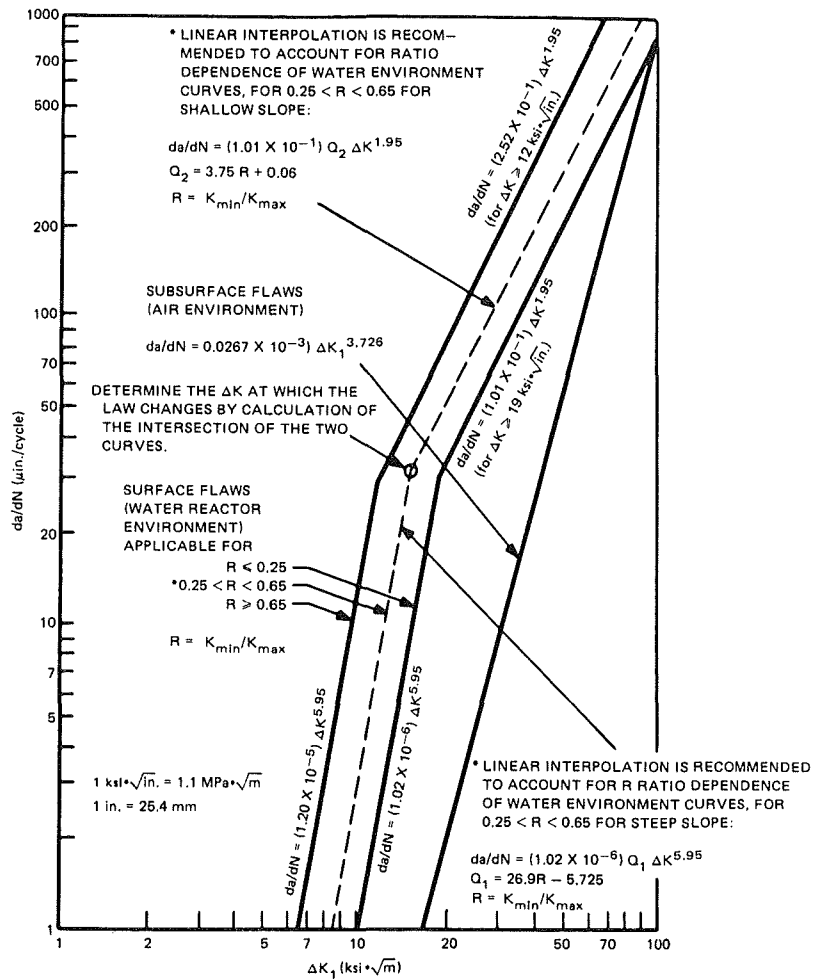
A533B and A508 Low-Alloy Steel: Theoretical and Observed da/dN versus ΔK Relationships



Comparison between theoretical and observed da/dN versus ΔK relationships for A533B and A508 steels with 0.01 to 0.025% S in 200 to 300 ppb O_2 water at 0.0167 Hz and $R = 0.7$.

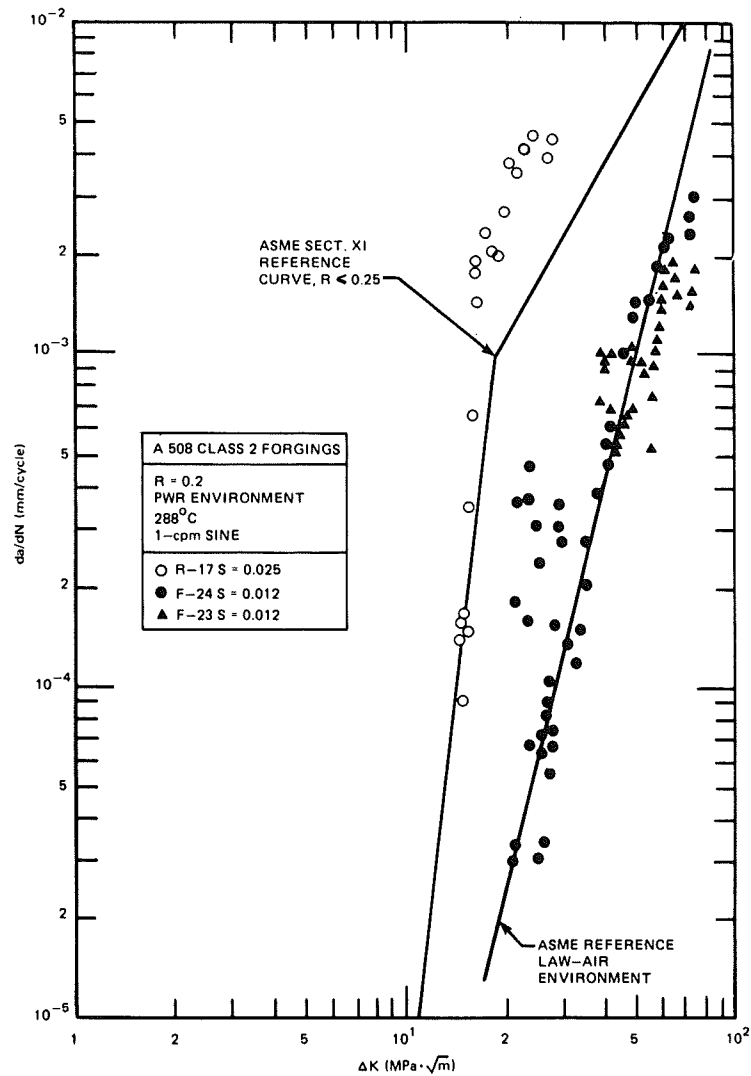
ASME Boiler and Pressure Vessel Code, 1986, Section XI,
Appendix A: Reference Fatigue Crack Growth Rate Laws for
Ferritic Pressure Vessel Steels

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Reference fatigue crack growth rate laws for ferritic pressure vessel steels.

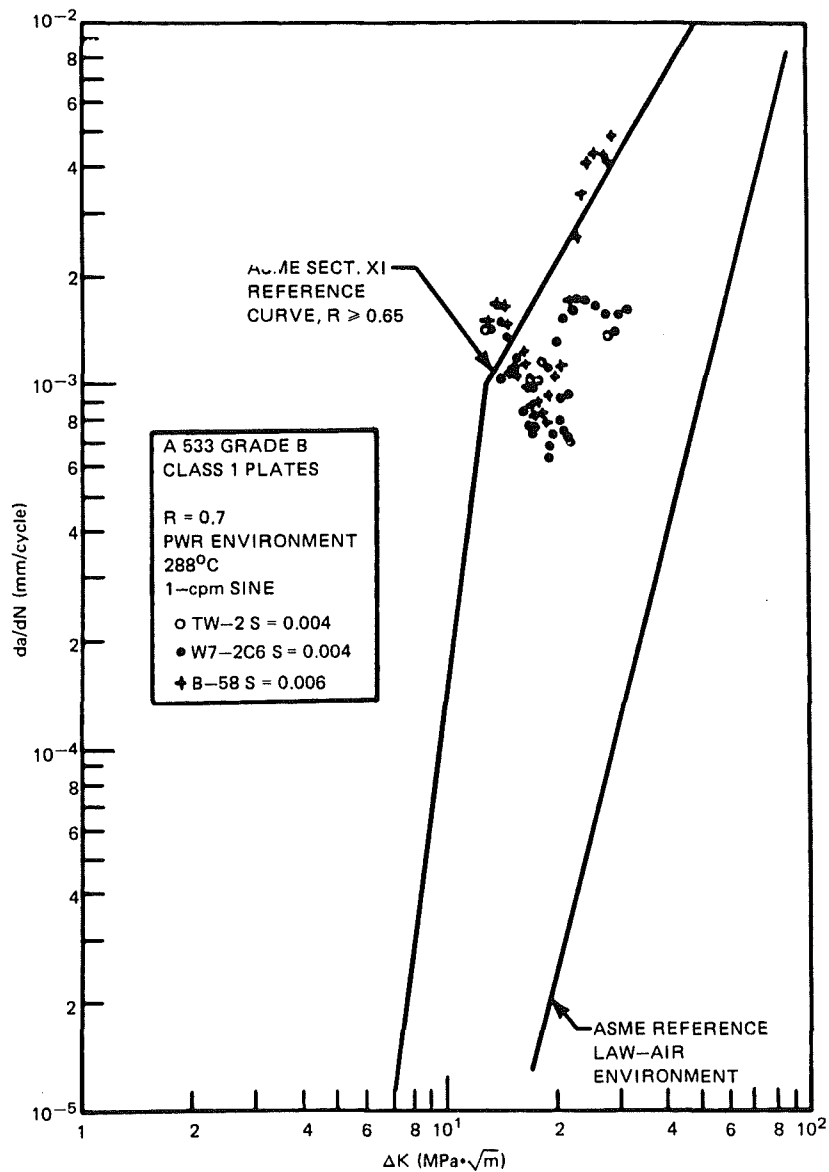
A508 Class 2 Forgings: Effect of Sulfur Content on Fatigue Crack Growth, in a PWR Environment



Effect of sulfur content of A508 Class 2 forgings on fatigue crack growth in a PWR environment at $R = 0.2$.

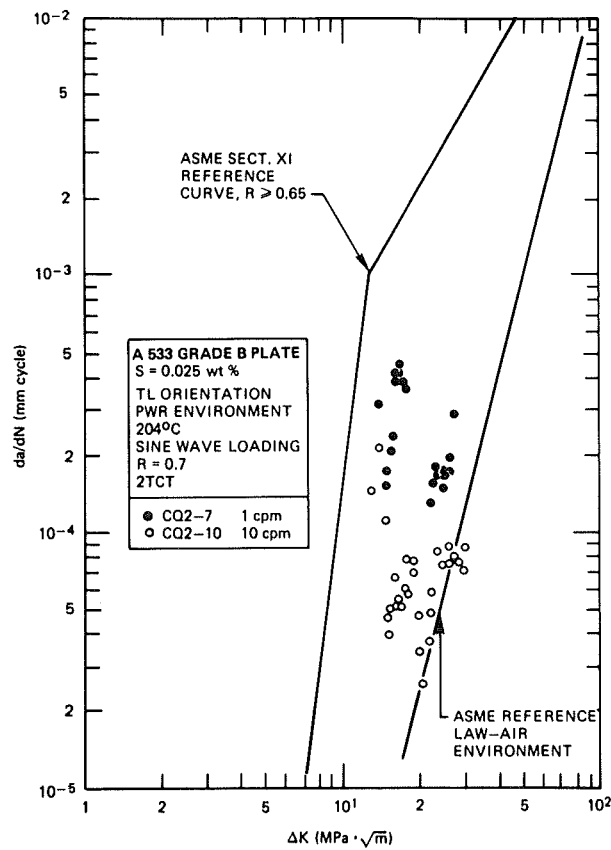
A533B Plate: Comparison of Fatigue Crack Growth Rates, in a PWR Environment

127



Comparison of fatigue crack growth rates for low-sulfur (A533 Grade B) plate tested at $R = 0.7$ in a PWR environment.

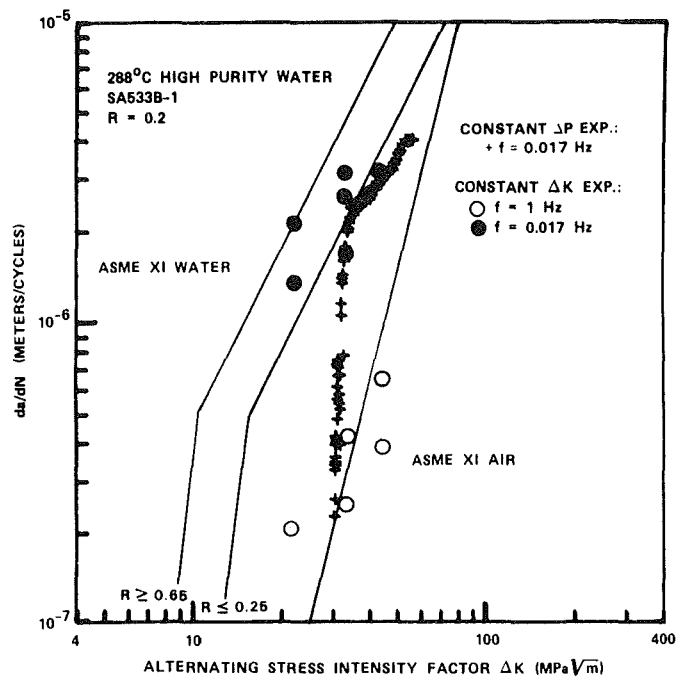
A533B Steel: Effect of Frequency on Fatigue Crack Growth, in a Simulated PWR Environment



Effect of frequency on fatigue crack growth for a high-sulfur heat of A533B steel at R = 0.7 at 204 °C in a simulated PWR environment.

SA533B-1 High-Strength Low-Alloy Steel: Fatigue Crack Growth Rates, in Deoxygenated High-Purity Water

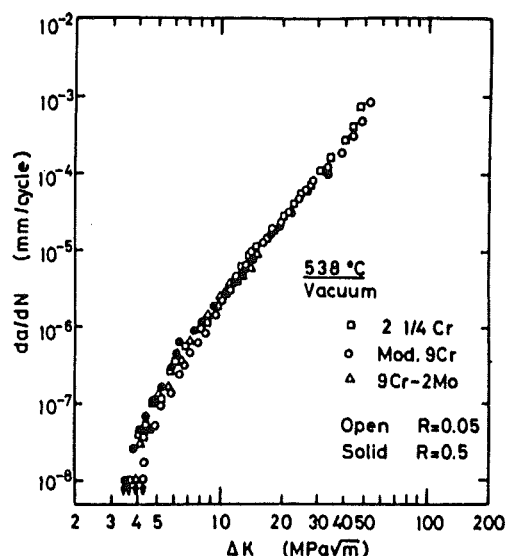
129



Fatigue crack growth rates for SA533B-1 steel in 288 °C deoxy-
genated high-purity water.

Note that the fatigue crack growth rates in high-purity water are
significantly higher than in a PWR water environment.

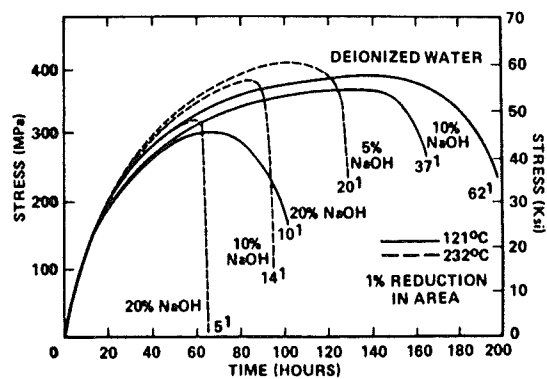
2.25Cr, Modified 9Cr, and 9Cr-2Mo Alloy Steels: Fatigue Crack Growth Rates, in Vacuum



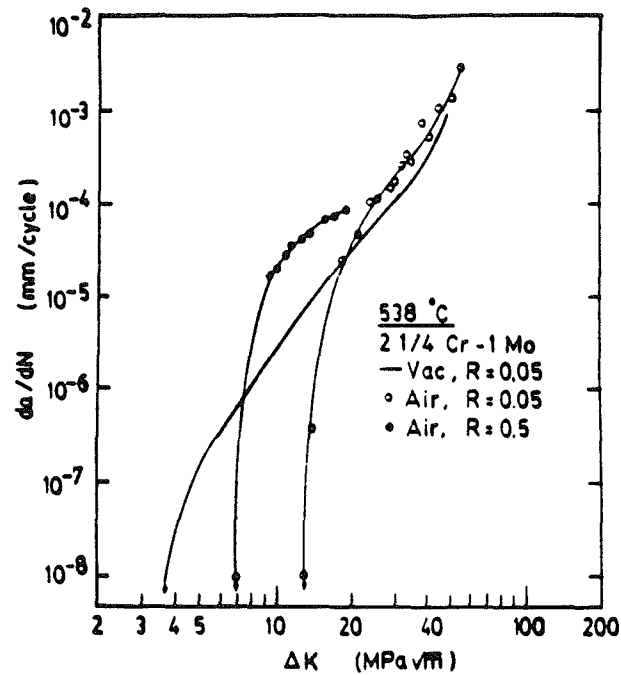
Rate of fatigue crack growth for three steels tested in vacuum at 538 °C at two R ratios.

Note that the rate of fatigue crack growth at 30 Hz in vacuum is independent of microstructure and of the R value. The absence of an R dependency indicates that crack closure is absent in these tests at elevated temperature. Additional tests at frequencies down to 0.3 Hz in vacuum show no effect of frequency, an indication that creep plays no role over the range of frequencies examined. In air, on the other hand, the rate of crack growth increases with decreasing frequency due to the deleterious effects of oxidation.

2.25Cr-1Mo Steel: Stress-Time Curves



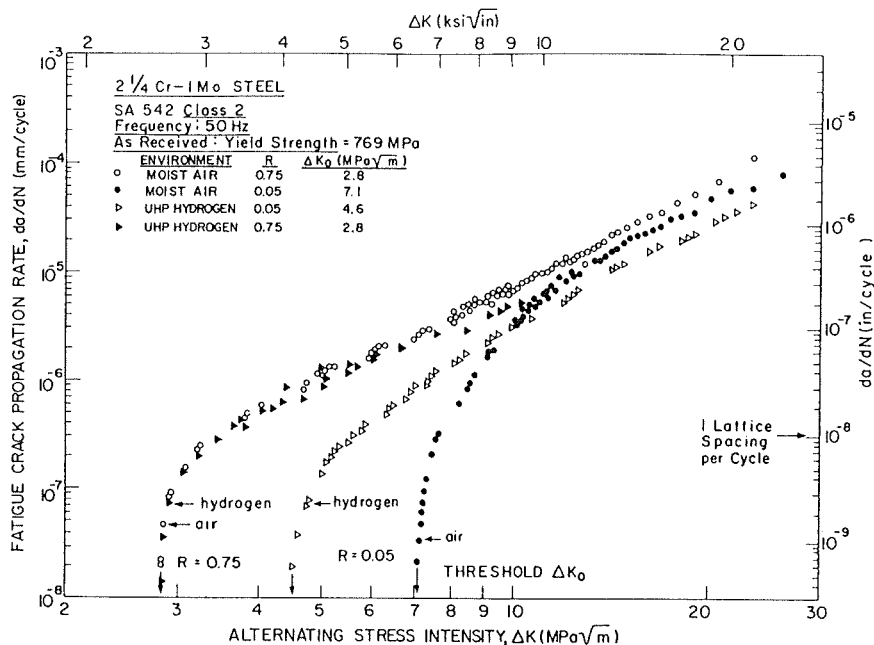
Stress-time curves for VAR 2.25Cr-1Mo steel strained at $4.4 \times 10^{-7}/s$ as a function of temperature and caustic concentration.



Fatigue crack growth rate of 2.25Cr-1Mo steel as a function of ΔK in air at 538 °C.

Note that the threshold in air is higher than in vacuum because of oxide-filling of the crack.

2.25Cr-Mo Alloy Steel: Fatigue Crack Propagation, in Air and in Hydrogen

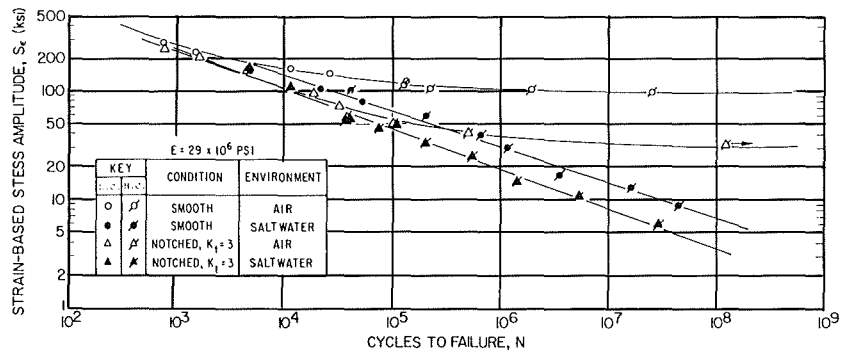


Fatigue crack propagation in martensitic 2.25Cr-Mo steel tested in moist air and dry hydrogen gas.

At low load ratios, near threshold, the fatigue crack growth rate in moist air is less than that in dry hydrogen due to the formation of excess corrosion deposits on the crack faces, which enhances crack closure and thereby reduces the effective range of the stress-intensity factor.

HP 9Ni-4Co-0.20C Alloy Steel: Flexural Fatigue Curves, in Air and in Salt Water

135

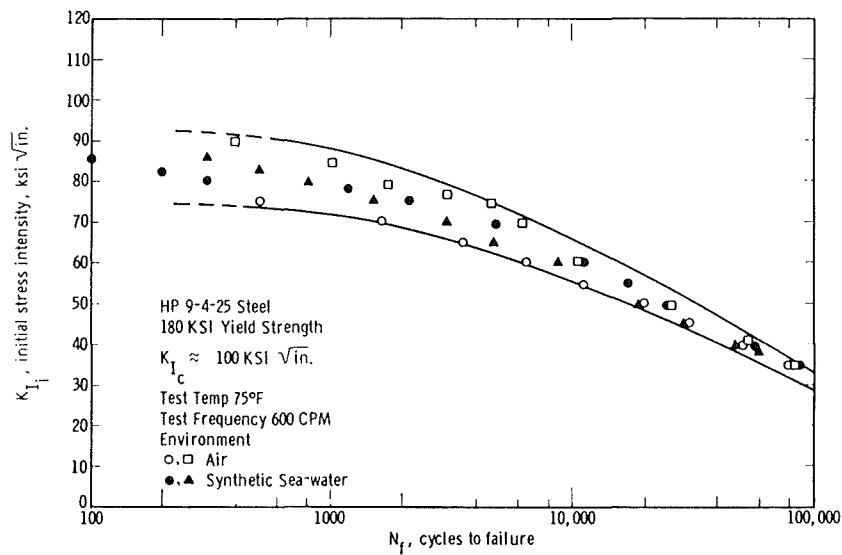


Flexural fatigue curves for HP 9Ni-4Co-0.20C steel.

The yield strength of this material was in the 180 to 200 ksi range.

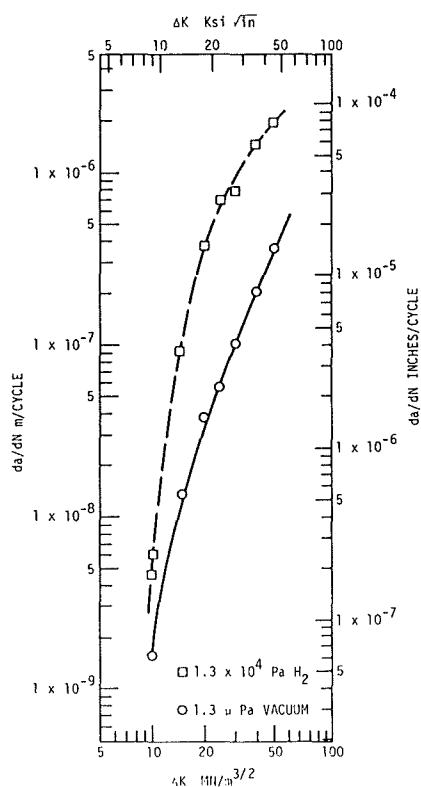
Source: R.C. Schwab and E.J. Czyryca, Effects of Notches and Saltwater Corrosion on the Flexural Fatigue Behavior of High-Strength Structural Alloys, in Effects of Environment and Complex Load History on Fatigue Life, STP 462, American Society for Testing and Materials, Philadelphia, 1970, p 203-216.

HP 9Ni-4Co-0.25C Alloy Steel: Effect of Environment on Fatigue Life



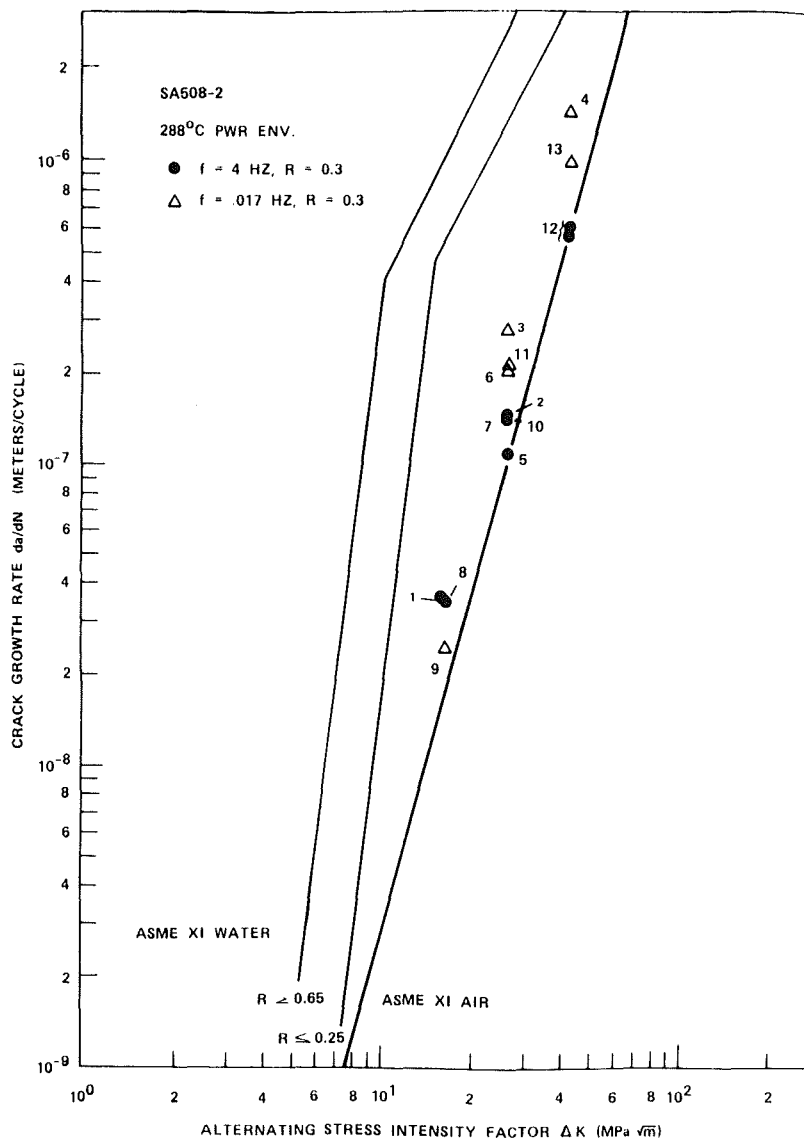
Effect of environment on fatigue life of HP 9Ni-4Co-0.25C alloy steel using 2-in. thick wedge opening load specimens.

Source: W.G. Clark, Jr. and E.T. Wessel, Influence of a Synthetic Seawater Environment on the Fracture Behavior of HP 9-4-25 and HP 9-4-20 Alloy Steels, in Materials Performance and the Deep Sea, American Society for Testing and Materials, Philadelphia, 1969, p 93-114.



Fatigue crack propagation rate of HP 9-4-20 steel in hydrogen and vacuum.

SA508-2 High-Strength Low-Alloy Steel: Fatigue Crack Growth Rates, in PWR Water

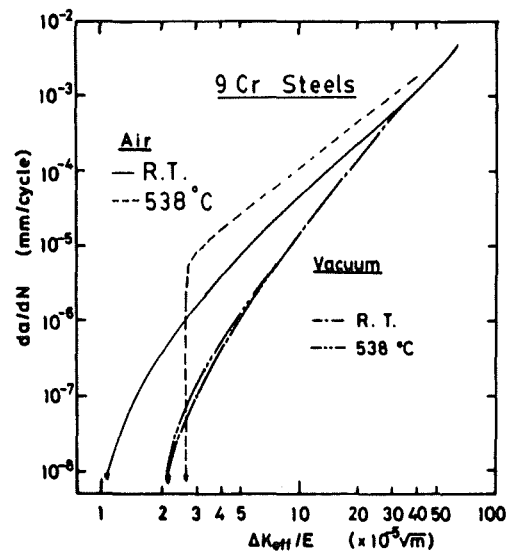


Fatigue crack growth rates for SA508-2 steel at 288 °C in PWR water.

The numbers on the figure refer to individual tests. The results are compared with the ASTM Boiler and Pressure Vessel Code, Section XI, upper bound fatigue crack growth data lines for reactor pressure vessel steels. Note that there is little effect of frequency on the rate of fatigue crack growth over the range of frequencies investigated.

9Cr Alloy Steels: Room-Temperature versus Elevated Temperature Fatigue Crack Propagation, in Air and in Vacuum

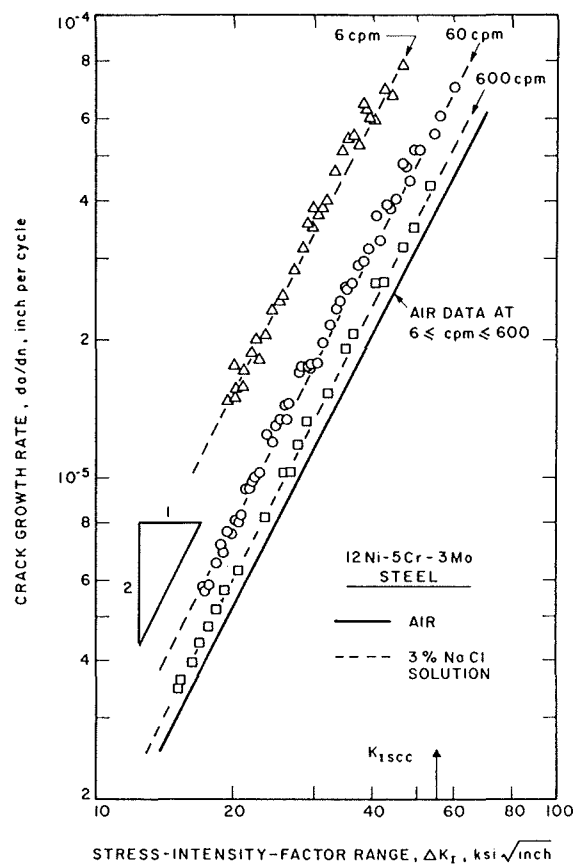
139



Comparison of room-temperature and elevated temperature fatigue crack propagation results in air and in vacuum at 30 Hz.

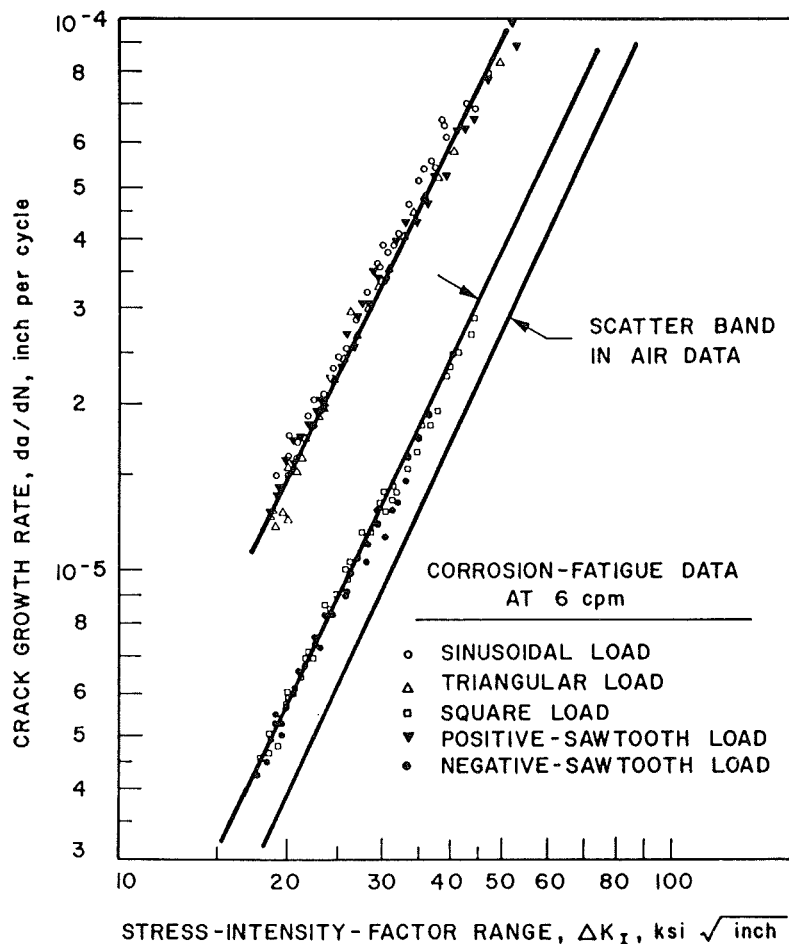
When the rate of fatigue crack growth is expressed as a function of $\Delta K_{eff}/E$, all results for vacuum fall on a single line. In air, however, because of the temperature dependence of the oxidation rate, such a correlation is not obtained. Note also that even after correcting for closure the threshold level in air at elevated temperature is higher than that in vacuum. This is interpreted to indicate that there is a strong oxide at the crack tip, which results in an increase in the threshold level. However, once the threshold level is exceeded at the elevated temperature, crack propagation in air occurs more rapidly than in vacuum.

12Ni-5Cr-3Mo Alloy Steel: Corrosion Fatigue Crack Growth, in Air and in Sodium Chloride



Corrosion fatigue crack growth data for 12Ni-5Cr-3Mo steel as a function of frequency in air and in 3% NaCl solution.

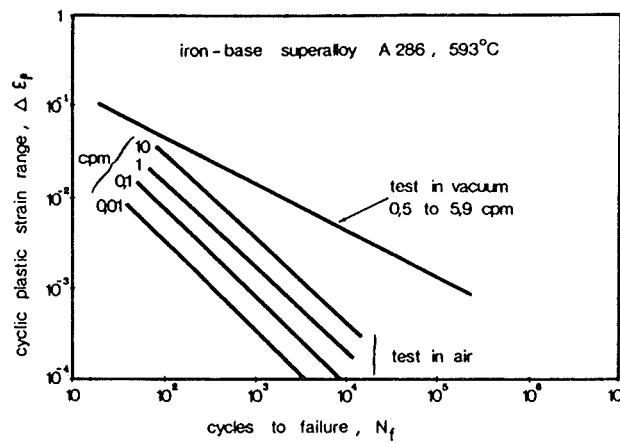
Source: E.J. Imhof and J.M. Barsom, Fatigue and Corrosion-Fatigue Crack Growth of 4340 Steel at Various Yield Strengths, in Progress in Flaw Growth and Fracture Toughness Testing, STP 536, American Society for Testing and Materials, Philadelphia, 1973, p 182-205.



Corrosion fatigue crack growth rates in 12Ni-5Cr-3Mo steel in 3% NaCl solution under various cyclic stress fluctuations with different stress-time profiles.

Note that the highest growth rates were found for those loading waveforms for which the strain rate during the rising portion of the cycle was low.

A286 Iron-Base Superalloy: Low-Cycle High-Temperature Fatigue Behavior, in Air and in Vacuum

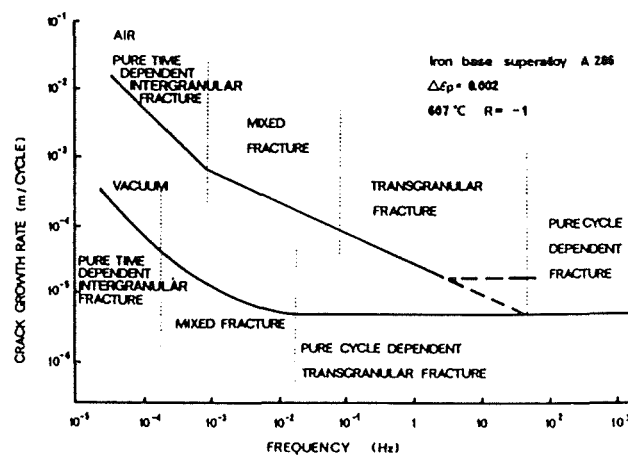


Low-cycle high-temperature fatigue behavior of A286, an iron-base superalloy, tested in air and in vacuum at 593 °C.

Composition: 26% Ni, 15% Cr, 1% Nb, 2% Ti, bal Fe. For this material, oxidation rather than creep is responsible for the decrease in fatigue resistance in air at 593 °C over the range of frequencies used.

A286 Iron-Base Superalloy: Effect of Frequency and Environment on the Rate of Fatigue Crack Growth

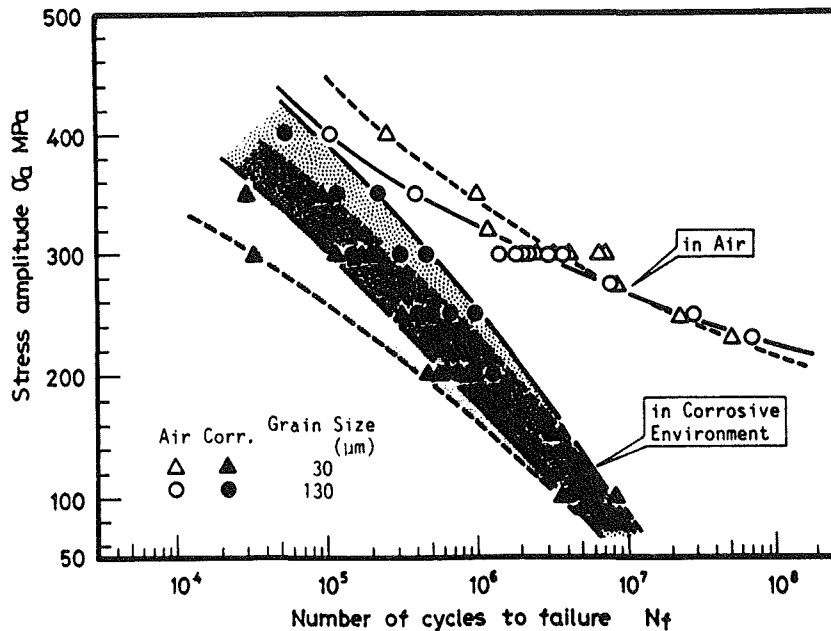
143



Overall view of the effect of frequency and environment on the rate of fatigue crack growth in an iron-base superalloy tested at 607°C .

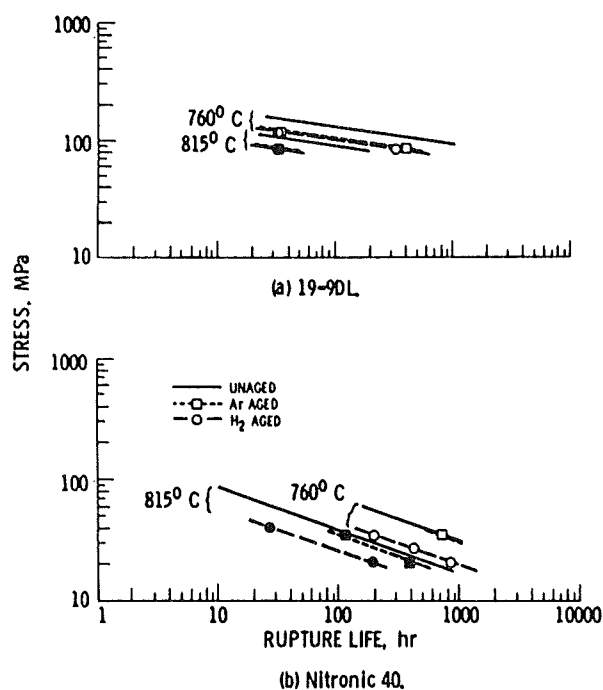
Composition: 26% Ni, 15% Cr, 1% Nb, 2% Ti, bal Fe. In this instance, at the lowest frequencies used, creep failures occurred.

Fe-42Ni-15Cr-3Mo Nickel Alloy: S/N Curves, in Air and in a Hot Corrosive Environment



S/N curves of Fe-42Ni-15Cr-3Mo specimens with different grain sizes at 800 °C in air and in a hot corrosive environment.

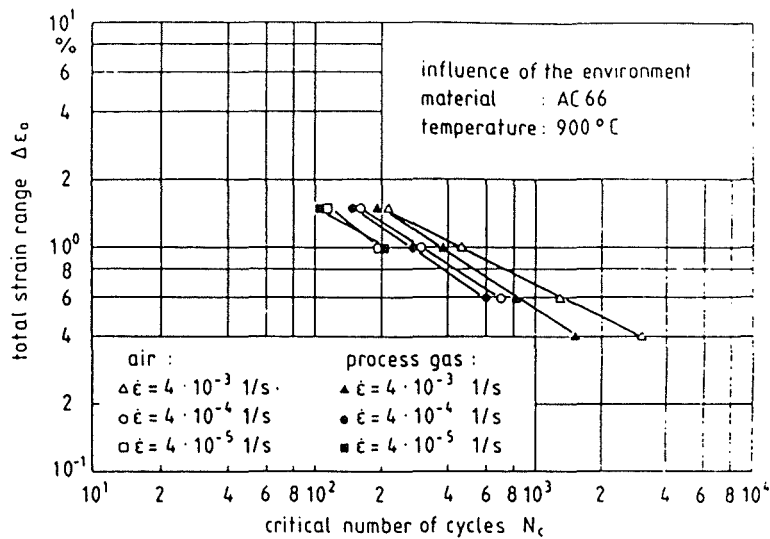
The fatigue tests were conducted at 800 °C using rotating bending fatigue machines operating at 1500 rpm. To simulate an actual hot corrosive environment, the specimens were coated with a synthetic salt mixture composed of 90% Na_2SO_4 plus 10% NaCl (melting point 785 °C). The amount of salty mixture applied was 0.4 kg/mm². In prolonged fatigue tests, the coating was reapplied at intervals of 3×10^5 cycles. Corrosion fatigue crack propagation was found to depend strongly on grain size. A smaller grain size results in an enhanced intergranular crack propagation rate because of the effect of mechanical fatigue and chemical corrosive damage concentrated at the grain boundaries. To improve the corrosion fatigue properties of superalloys, the grain size has to be large enough to avoid intergranular fracture.



Effect of hydrogen aging on stress-rupture behavior of candidate heater tube alloys.

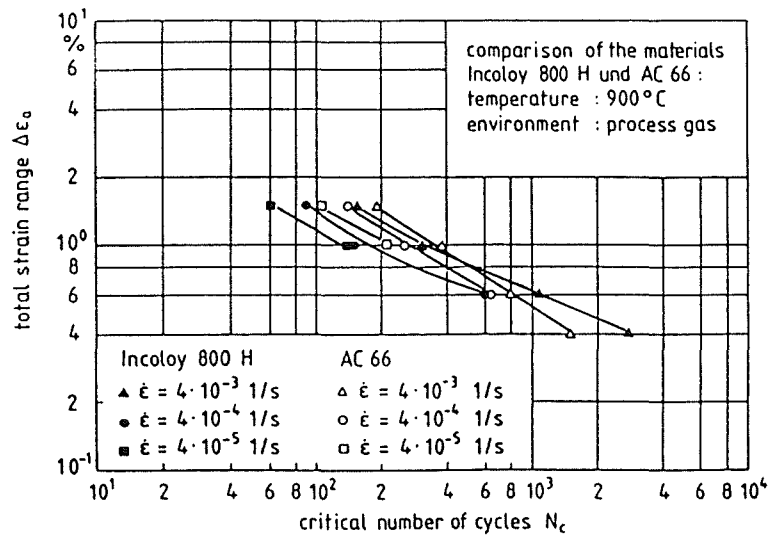
19-9DL and Nitronic 40 are iron-base alloys.

Alloy AC 66: Fatigue Life, in Air and in Process Gas



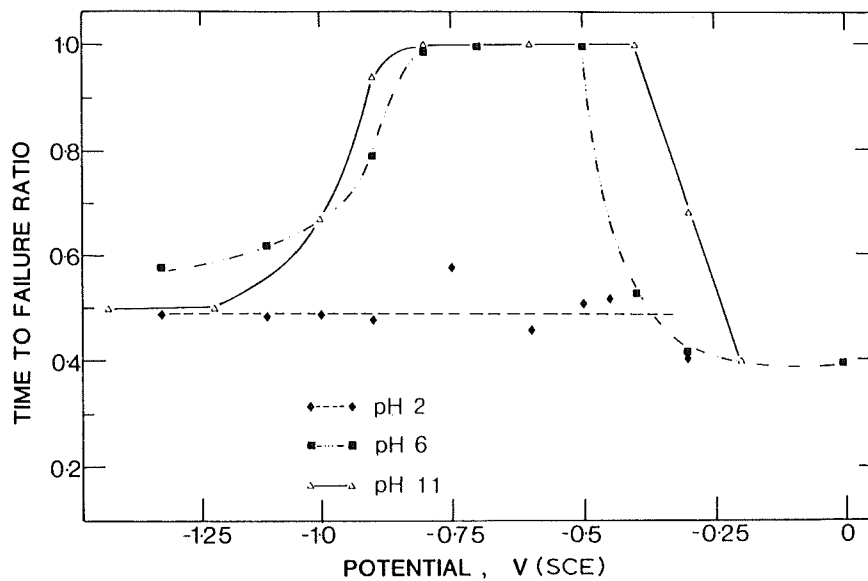
Fatigue life of AC 66 at 900 °C in air and in process gas.

AC 66: X5NiCrCeNb 3227. AC 66 is an iron-base superalloy. The reactor in a coal gasification plant operates at 900 °C and has to be supplied with the required process heat by a tubular heat exchanger. The material for the heat exchanger requires a certain resistance to corrosion fatigue—among other considerations, process heat by a tubular heat exchanger.



Source: K.-T. Rie, High Temperature Low Cycle Fatigue in Coal Gasification Environment, in Low Cycle Fatigue and Elasto-Plastic Behaviour of Materials, K.-T. Rie, Ed., Elsevier Applied Science Publishers, New York, 1987, p 389-394.

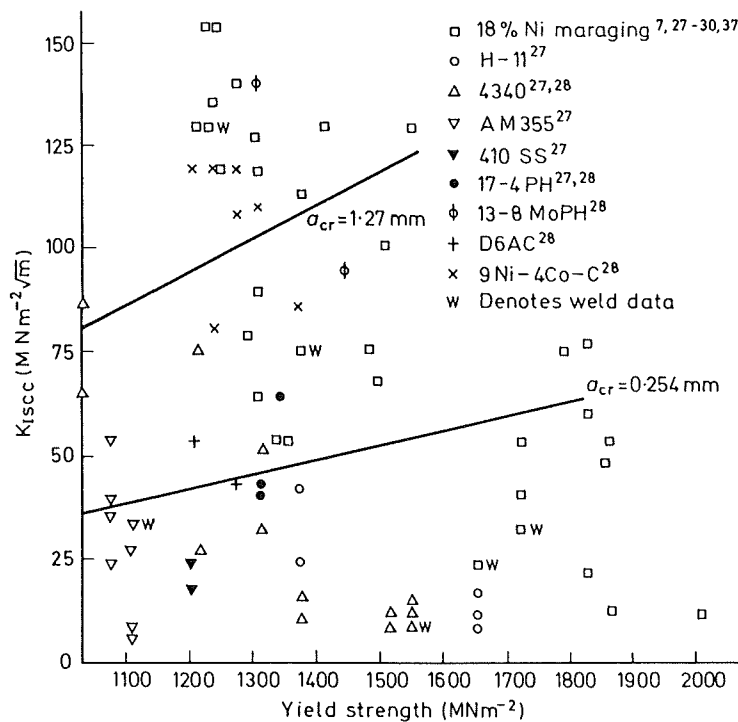
Maraging Steels



Time-to-failure in slow strain rate tests, normalized with respect to time-to-failure in air, as a function of applied potential, for maraging steel specimens tested in 3.5% sodium chloride at various initial pH values.

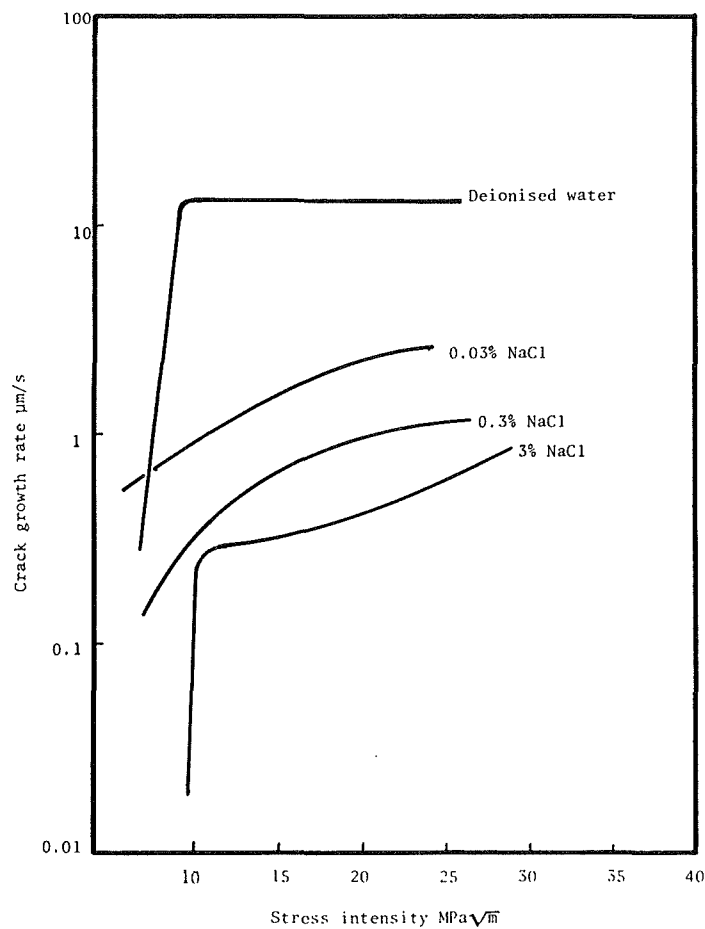
In plain specimens, ductile failures occurred at intermediate potentials in solutions with pH values of 6 or 11. At pH 2, environment-related failure occurred independent of potential over the range studied. However, when precracked specimens were tested in the pH 11 solution at those potentials (-0.5 and -0.7), which did not promote cracking in plain specimens, stress-corrosion cracking occurred, as it did at open circuit potential with precracked specimens at lower pH values. The precracks are probably important in relation to environmental effects, such as changes in pH, oxygen concentration, or potential variations within the crack enclave.

18Ni Maraging Steel: Comparison of K_{ISCC} as a Function of Yield Strength



Comparison of K_{ISCC} as a function of the yield strength of 18Ni maraging and other high-strength steels.

Maraging steels generally display superior resistance in terms of K_{ISCC} level in comparison to other high-strength steels. The resistance of maraging steels to stress-corrosion cracking generally follows fracture toughness. Processing techniques such as vacuum melting, proper hot working, and keeping residual impurity levels low also improve resistance to stress-corrosion cracking. Surface treatments such as painting or shot peening have been used to increase resistance. Cathodic protection can also be used, but must be employed carefully to avoid hydrogen charging that could initiate cracking.



Influence of chloride concentration and stress-intensity on the rate of stress-corrosion crack growth in a Fe-20Co-15Cr-5Mo maraging steel.

These results are for precracked wedge opening loaded specimens. The material had been age hardened by heat treatment at 500 °C for 38 h, which provided a favorable combination of fracture toughness and tensile strength.

Maraging Steel: Relationship Between Yield Strength and Stress-Corrosion Cracking Velocity

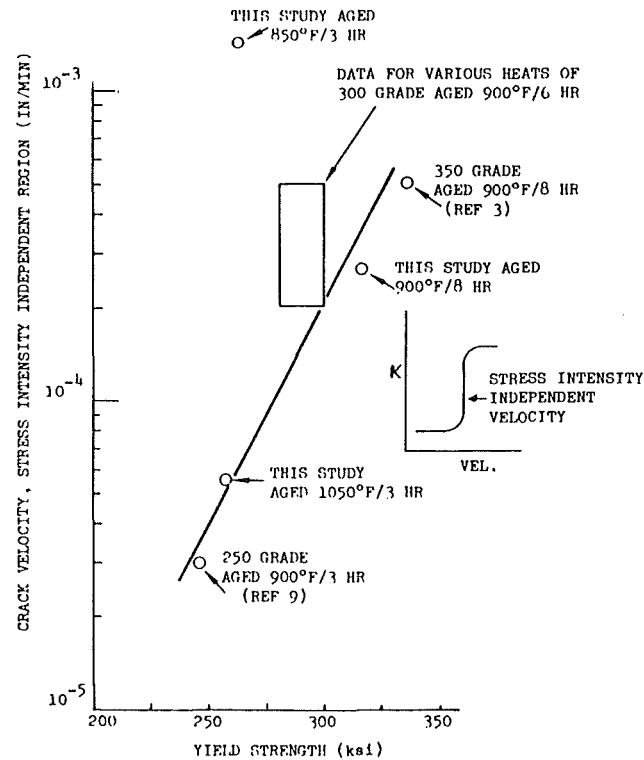
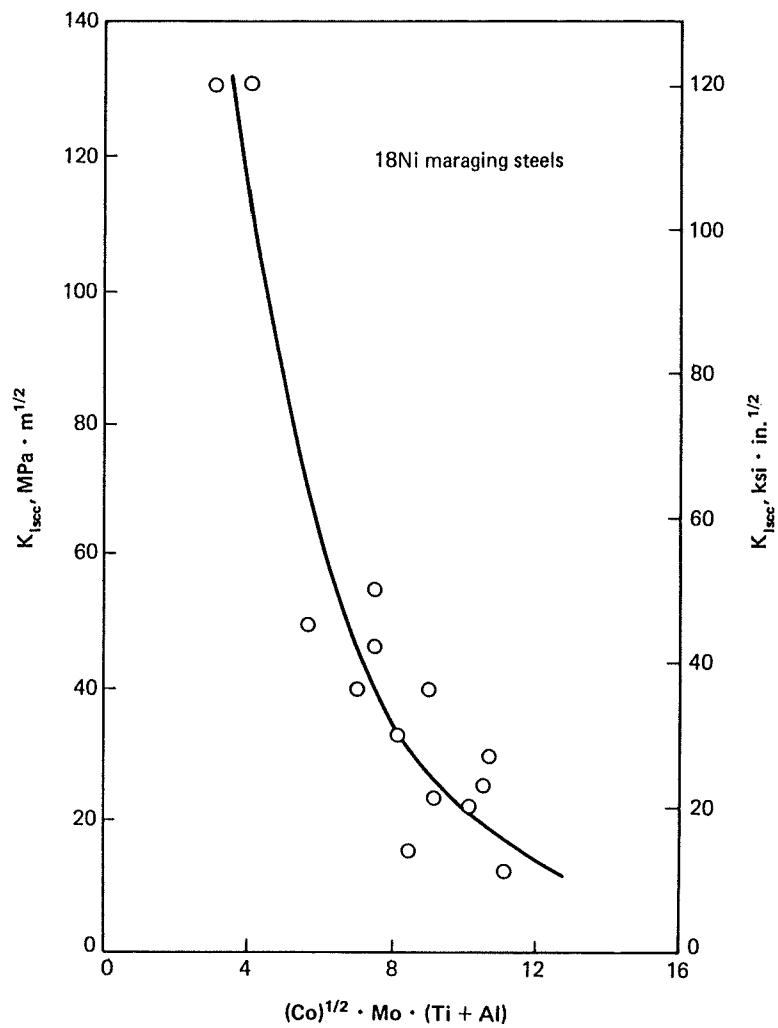


Fig. 10—Relationship between yield strength and stress corrosion crack velocity.

Relationship between yield strength and plateau stress-corrosion cracking velocity.

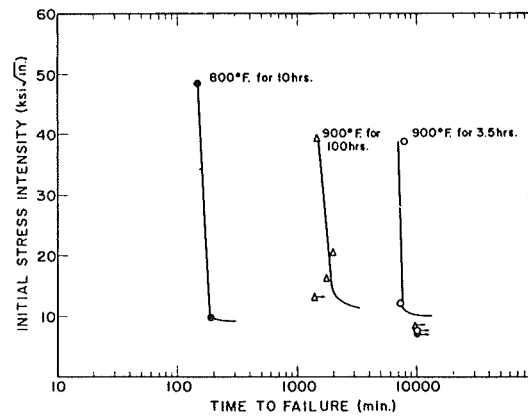
The maraging steel investigated had a nominal composition of 16.3Ni-12.87Co-4.98Mo-0.78Ti. The specimens were precracked Charpy bars, and they were loaded as cantilever beams. For stress-corrosion testing, a supply of 3.5% aqueous sodium chloride was dripped into the notch.



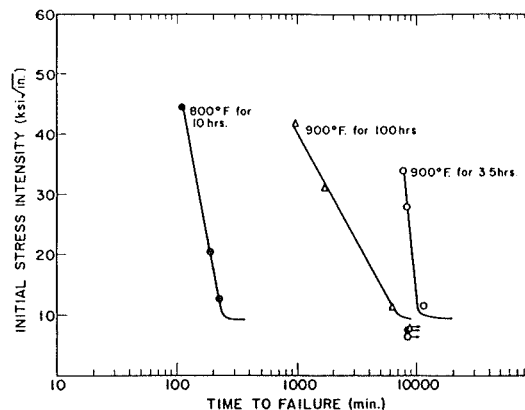
Effect of alloying element parameter on K_{Isc} for 18Ni maraging steels in an aqueous solution of sodium chloride.

Source: R.H. Van Stone and A.W. Gunderson, Fracture Properties of Carbon and Alloy Steels, in Application of Fracture Mechanics for Selection of Metallic Structural Materials, J.E. Campbell, W.W. Gerberich, and J.H. Underwood, Ed., American Society for Metals, Metals Park, 1982, p 41-104.

Maraging Steel: Effect of Aging Treatment

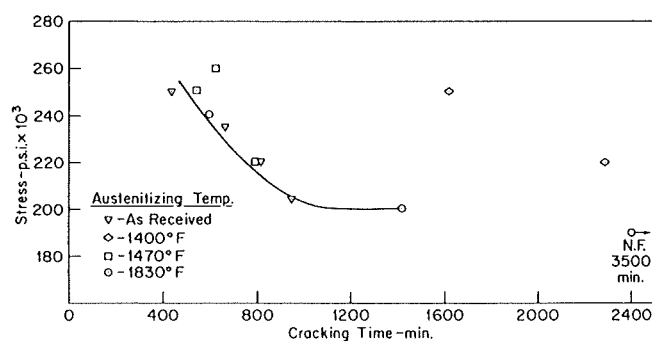
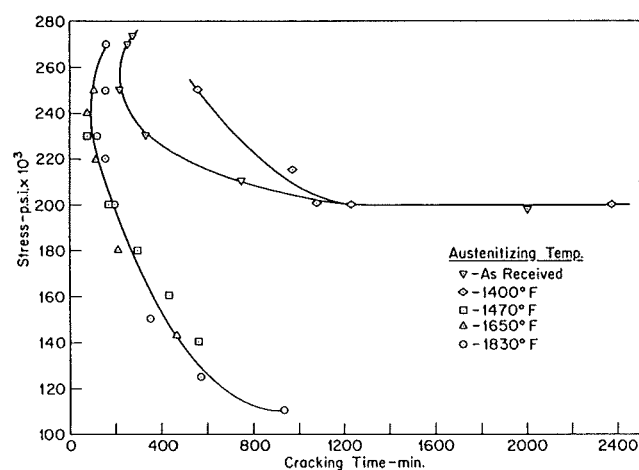


Effect of aging treatment on stress-corrosion cracking susceptibility in 3% sodium chloride at pH 1.7 of 18% Ni maraging steel austenitized 2 h at 1500 °F.



Effect of aging treatment on stress-corrosion cracking susceptibility in 3% sodium chloride at pH 6.3 of 18% Ni maraging steel austenitized 2 h at 1500 °F.

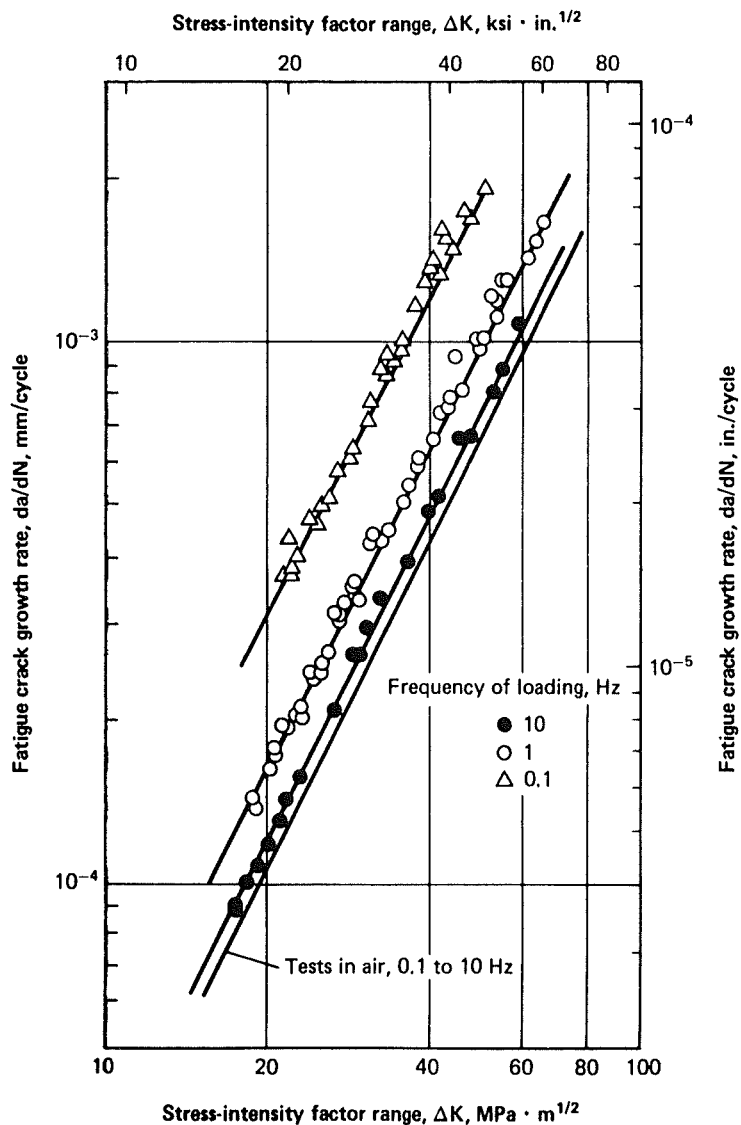
The stress-corrosion test results were obtained with cantilever beam specimens that contained a V-notch 0.22 in. deep and with a root radius of less than 0.01 in. The specimens were also side grooved on each side to a depth of 5% of the width. Specimens were heat treated to a hardness of 52 to 53 HRC. After heat treatment, the specimens were fatigue precracked. For pH levels in the range 1.7 to 6.3, the K_{ISCC} value was about 10 to 15 ksi-in.^{1/2} and independent of aging treatment. The time-to-failure was relatively insensitive to stress-intensity, but very dependent on aging treatment. The crack path was along prior austenitic grain boundaries, and the results were consistent with a hydrogen embrittlement mechanism.



Effects of stress and austenitizing temperature on the total time-to-failure for one heat of an 18% Ni maraging steel.

The specimens were unnotched and were tested in 0.6N sodium chloride at pH 2.2. The advantage of keeping the austenitizing temperature as low as possible is illustrated. Numerous cracks were initiated, and in the most susceptible conditions, the cracks followed the original austenite grain boundaries. When heat treated to improve resistance to stress-corrosion cracking, the cracks became appreciably less branched and tended to become transgranular. Small particles, possibly titanium carbide, were detected along the prior austenitic grain boundaries, and these particles were most readily detectable in the microstructures most susceptible to stress-corrosion cracking.

12Ni-5Cr-3Mo Maraging Steel: Effect of Cyclic Frequency on Corrosion Fatigue Crack Growth

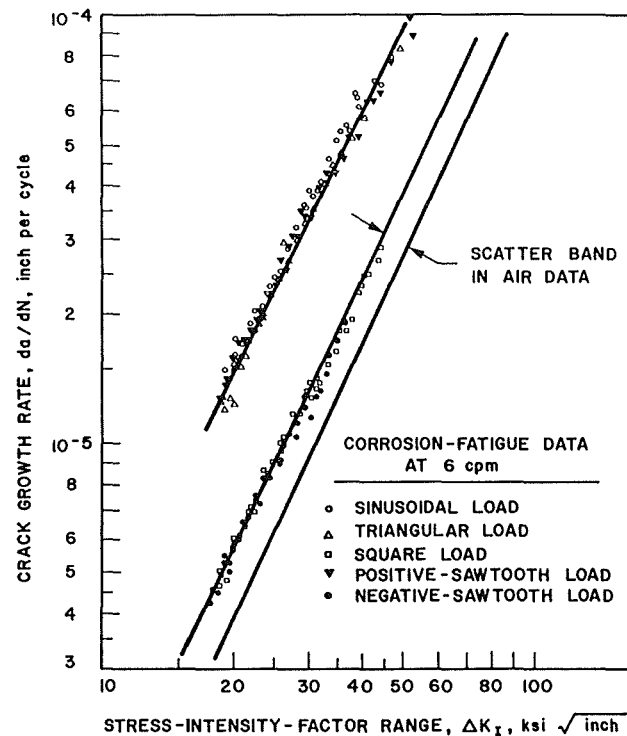


Effect of cyclic frequency on the corrosion fatigue crack growth rates for 12Ni-5Cr-3Mo maraging steel. The steel was tested in air and in a 3% aqueous solution of sodium chloride with sinusoidal loading.

Source: R.H. Van Stone and A.W. Gunderson, Fracture Properties of Carbon and Alloy Steels, in Application of Fracture Mechanics for Selection of Metallic Structural Materials, J.E. Campbell, W.W. Gerberich, and J.H. Underwood, Ed., American Society for Metals, Metals Park, 1982, p 41-104.

12Ni-5Cr-3Mo Maraging Steel: Effect of Shape of Loading Wave Form on Corrosion Fatigue Crack Growth

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Effect of shape of loading wave form on the corrosion fatigue crack growth rates for 12Ni-5Cr-3Mo maraging steel.

The steel was tested in air and in a 3% aqueous solution of sodium chloride. Wave forms that had a relatively slow rise in load during tensile loading were associated with the highest rates of crack growth.

STAINLESS STEELS

STAINLESS STEELS

All austenitic grades of stainless steels, especially AISI types 304 and 316, are susceptible to stress-corrosion cracking (SCC) in chloride environments, and this susceptibility is increased by the presence of oxygen and by elevated temperatures. The cracking path is usually transgranular, but grain boundary sensitization due to the precipitation of carbide, nitride, and intermetallic phases leads to cracking in the chromium depleted matrix adjacent to sensitized grain boundaries. On the other hand, most ferritic and duplex stainless steels are either immune or highly resistant to SCC, although the ferritic AISI types 430, 434, and 446 can also be sensitized. The highly alloyed austenitic grades are more resistant to sodium chloride (NaCl) solutions, but crack readily in MgCl_2 solutions. Sour environments can result in sulfide stress cracking (SSC) in susceptible alloys. The resistance to SSC increases with decreased hardness. Duplex and ferritic grades have been used for their improved resistance to SSC. High-strength martensitic grades are susceptible to hydrogen assisted cracking.

Although some localized pitting or crevice corrosion probably precedes SCC, the amount of pitting or crevice attack may be so small as to be undetectable. Stress corrosion is difficult to detect while in progress, and even when pervasive, can lead to rapid catastrophic failures of pressurized equipment. It is difficult to alleviate the environmental conditions that lead to SCC as the level of chlorides required to produce stress corrosion is very low. In operation, there can be evaporative concentration of chlorides or a concentration in the surface film on a heat-rejecting surface. Temperature is often a process parameter, as in the case of a heat exchanger. Although tensile stress is one parameter that might be controlled, the residual stresses associated with fabrication, welding, or thermal cycling, rather than design stresses, are often responsible for SCC, and even stress-relieving heat treatments do not completely eliminate these residual stresses.

Austenitic Stainless Steel: Stress-Corrosion Cracking Behavior in Boiling Magnesium Chloride

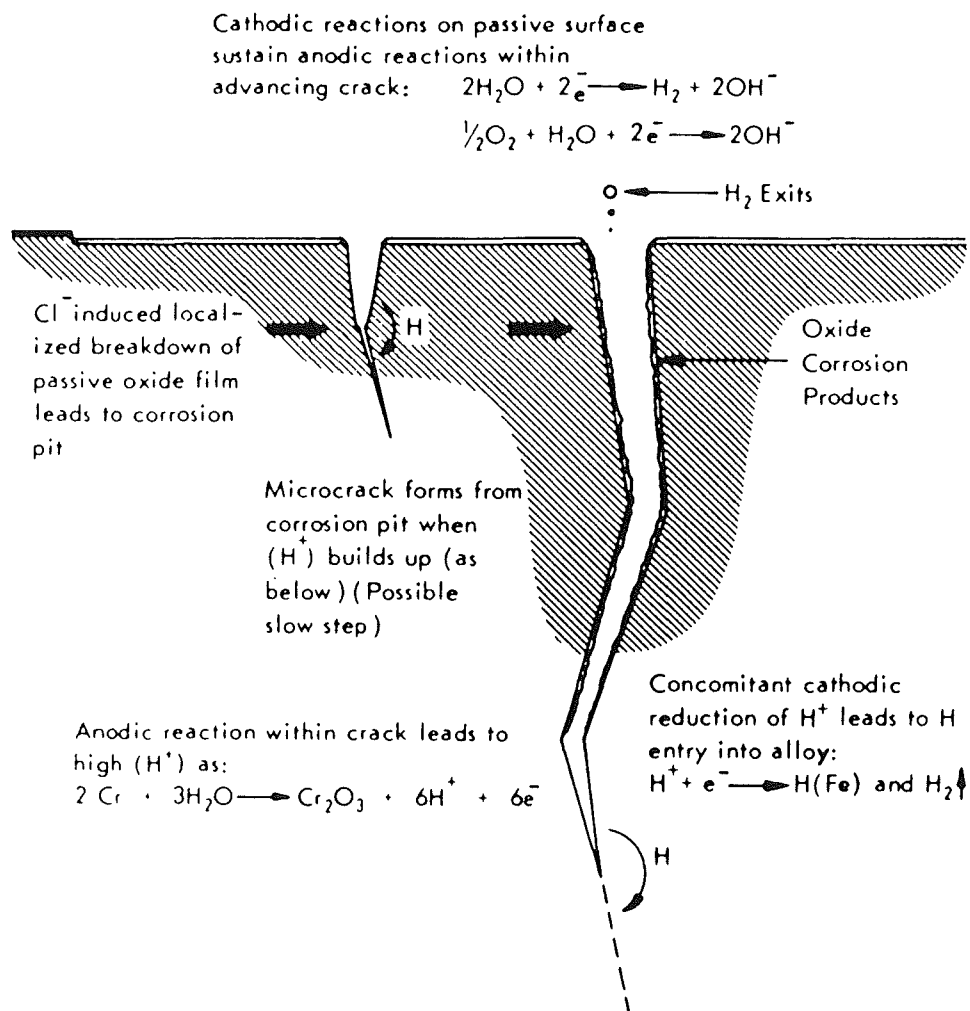
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**Compositions of Standard Grades of Wrought Stainless and
Heat-Resisting Steels (AISI, 1959)**

AISI type	Nominal composition, %					Other(a)
	C	Mn, max	Si, max	Cr	Ni	
Austenitic Steels						
201	0.15 max	7.50(b)	1.00	16.00 to 18.00	3.50 to 5.50	0.25 max N
202	0.15 max	10.00(c)	1.00	17.00 to 19.00	4.00 to 6.00	0.25 max N
301	0.15 max	2.00	1.00	16.00 to 18.00	6.00 to 8.00	
302	0.15 max	2.00	1.00	17.00 to 19.00	8.00 to 10.00	
302B	0.15 max	2.00	3.00(d)	17.00 to 19.00	8.00 to 10.00	
303	0.15 max	2.00	1.00	17.00 to 19.00	8.00 to 10.00	0.15 min S
303(Se)	0.15 max	2.00	1.00	17.00 to 19.00	8.00 to 10.00	0.15 min Se
304	0.08 max	2.00	1.00	18.00 to 20.00	8.00 to 12.00	
304L	0.03 max	2.00	1.00	18.00 to 20.00	8.00 to 12.00	
305	0.12 max	2.00	1.00	17.00 to 19.00	10.00 to 13.00	
308	0.08 max	2.00	1.00	19.00 to 21.00	10.00 to 12.00	
309	0.20 max	2.00	1.00	22.00 to 24.00	12.00 to 15.00	
309S	0.08 max	2.00	1.00	22.00 to 24.00	12.00 to 15.00	
310	0.25 max	2.00	1.50	24.00 to 26.00	19.00 to 22.00	
310S	0.08 max	2.00	1.50	24.00 to 26.00	19.00 to 22.00	
314	0.25 max	2.00	3.00(e)	23.00 to 26.00	19.00 to 22.00	
316	0.08 max	2.00	1.00	16.00 to 18.00	10.00 to 14.00	2.00 to 3.00 Mo
316L	0.03 max	2.00	1.00	16.00 to 18.00	10.00 to 14.00	2.00 to 3.00 Mo
317	0.08 max	2.00	1.00	18.00 to 20.00	11.00 to 15.00	3.00 to 4.00 Mo
321	0.08 max	2.00	1.00	17.00 to 19.00	9.00 to 12.00	5 × C min Ti
347	0.08 max	2.00	1.00	17.00 to 19.00	9.00 to 13.00	10 × C min Cb-Ta
348	0.08 max	2.00	1.00	17.00 to 19.00	9.00 to 13.00	10 × C min Cb-Ta 0.10 max Ta
Martensitic Steels						
403	0.15 max	1.00	0.50	11.50 to 13.00		
410	0.15 max	1.00	1.00	11.50 to 13.50		
414	0.15 max	1.00	1.00	11.50 to 13.50	1.25 to 2.50	
416	0.15 max	1.25	1.00	12.00 to 14.00		0.15 min S
416(Se)	0.15 max	1.25	1.00	12.00 to 14.00		0.15 min Se
420	0.15 min	1.00	1.00	12.00 to 14.00		
431	0.20 max	1.00	1.00	15.00 to 17.00	1.25 to 2.50	
440A	0.60 to 0.75	1.00	1.00	16.00 to 18.00		0.75 max Mo
440B	0.75 to 0.95	1.00	1.00	16.00 to 18.00		0.75 max Mo
440C	0.95 to 1.20	1.00	1.00	16.00 to 18.00		0.75 max Mo
501	0.10 min	1.00	1.00	4.00 to 6.00		0.40 to 0.65 Mo
502	0.10 max	1.00	1.00	4.00 to 6.00		0.40 to 0.65 Mo
Ferritic Steels						
405	0.08 max	1.00	1.00	11.50 to 14.50		0.10 to 0.30 Al
430	0.12 max	1.00	1.00	14.00 to 18.00		
430F	0.12 max	1.25	1.00	14.00 to 18.00		0.15 min S
430F(Se) ..	0.12 max	1.25	1.00	14.00 to 18.00		0.15 min Se
446	0.20 max	1.50	1.00	23.00 to 27.00		0.25 max N

(a) Other elements in addition to those shown below are as follows: Phosphorus is 0.06% max in types 201, 202, 416, 416(Se), 430F, 430F(Se); 0.045% max in types 301, 302, 302B, 304, 304L, 305, 308, 309, 309S, 310, 310S, 314, 316, 316L, 317, 321, 347, 348; 0.20% max in types 303 and 303(Se). Sulfur is 0.030% max in types 201, 202, 301, 302, 302B, 304, 304L, 305, 308, 309, 309S, 310, 310S, 314, 316, 316L, 317, 321, 347, 348, 403, 405, 410, 414, 420, 430, 431, 440A, 440B, 440C, 446, 501 and 502; 0.15% min in types 303, 416 and 430F. (b) Mn range, 5.50 to 7.50. (c) Mn range, 7.50 to 10.00 (d) Si range, 2.00 to 3.00. (e) Si range, 1.50 to 3.00.

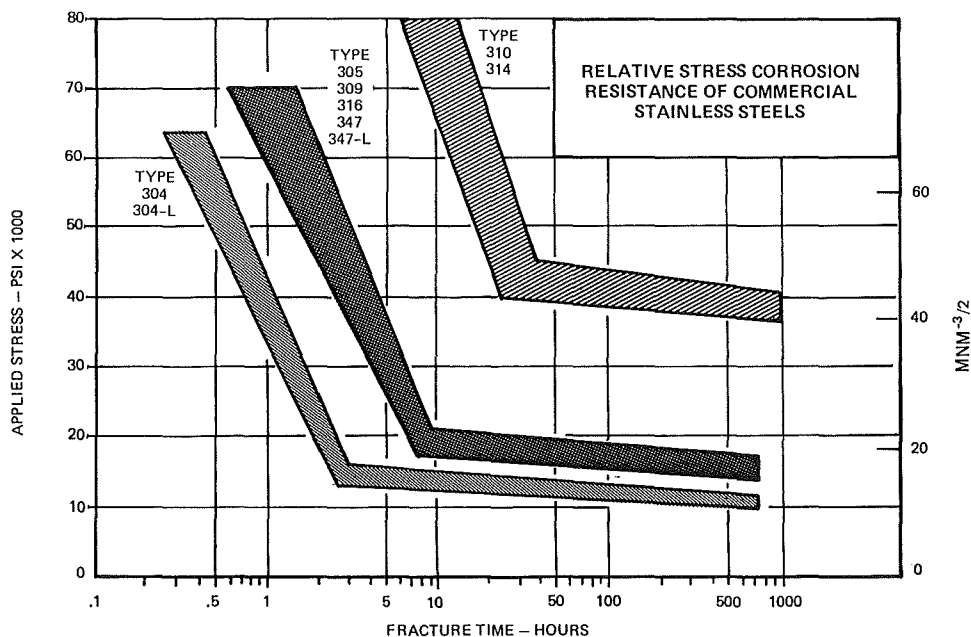
Austenitic Stainless Steel: Principal Factors Involved in Initiation and Propagation of Stress-Corrosion Cracks



Schematic representation of the principal factors involved in the initiation and propagation of stress-corrosion cracks in austenitic stainless steels.

Austenitic Stainless Steel: Stress-Corrosion Cracking Behavior in Boiling Magnesium Chloride (Continued)

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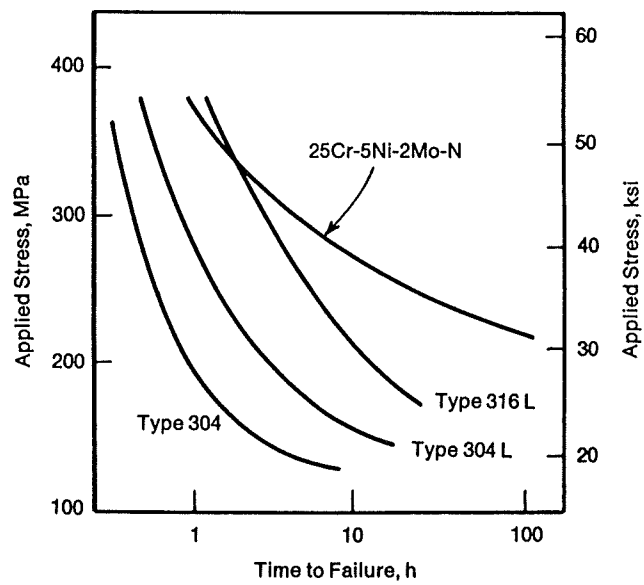
Relative stress-corrosion cracking behavior of major types of austenitic stainless steels in boiling magnesium chloride.

Much accelerated chloride cracking treating has been carried out in boiling magnesium chloride solution. Although testing in boiling magnesium chloride solution can be performed with various concentrations of magnesium chloride, ASTM G 36 specifies a test solution maintained at a constant boiling temperature of 155.0 ± 1.0 °C, that is approximately 45% magnesium chloride. However, most of the stress-corrosion cracking data on austenitic stainless steels have been obtained by using a boiling solution of 42% magnesium chloride (154 °C), and much current testing is done at the lower concentration.

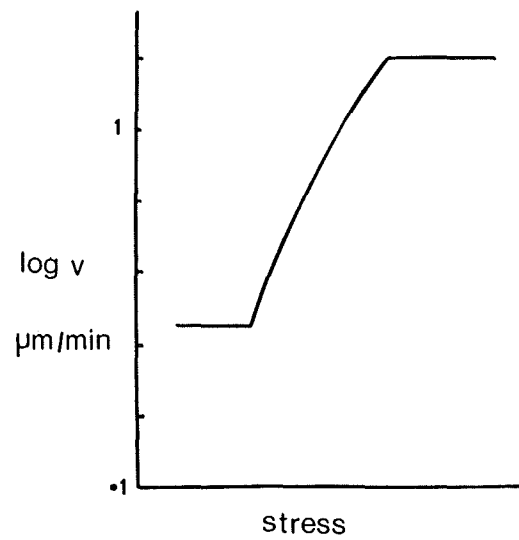
The accompanying table lists the compositions of standard grades of wrought stainless and heat-resisting steels. Note that types 310 and 314 contain 19 to 22% nickel, which makes them more resistant to chloride stress-corrosion cracking. An increase in nickel content, and the addition of molybdenum, titanium, and niobium to take up carbon and prevent chromium carbide formation, can also prove to be beneficial.

Fully ferritic stainless steels are highly resistant to stress-corrosion cracking in chloride and caustic environments that cause austenitic steels to crack. However, small additions of nickel or copper to ferritic steels may render them susceptible to stress-corrosion cracking in severe environments.

Stainless Steel: Stress-Corrosion Cracking in Boiling Magnesium Chloride

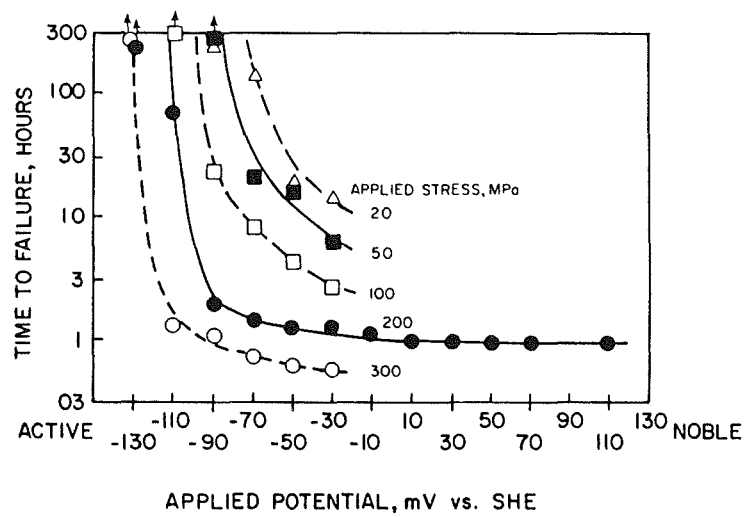


Stress-corrosion cracking resistance of stainless steels in boiling 42% magnesium chloride.

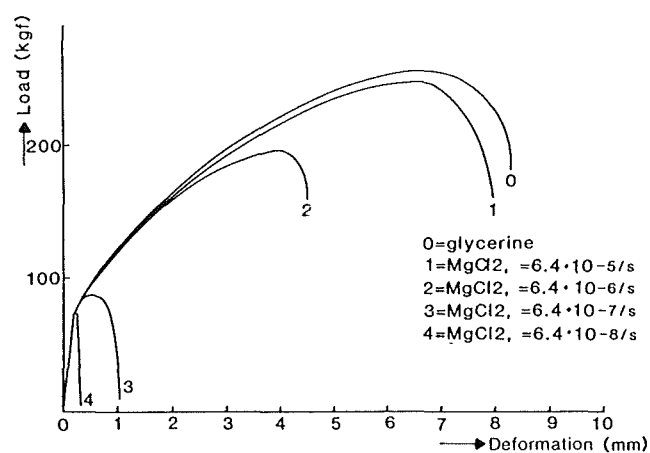


Stress-corrosion crack velocity in a type 304 stainless steel in magnesium chloride solution at 157.5°C.

Type 304 Stainless Steel: Relationship Between Applied Potential, Applied Stress, and Time-to-Failure

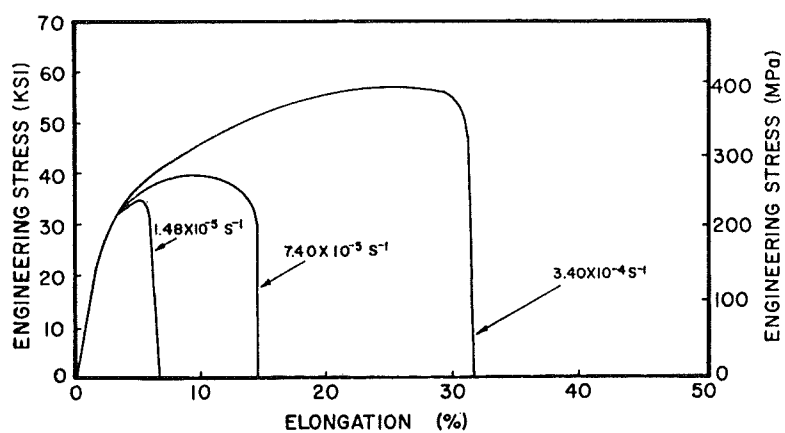


Relationship between applied potential, applied stress, and time-to-failure of solution annealed type 304 stainless steel in a 42% magnesium chloride solution at 144 °C.



Slow strain rate test results on type 304 stainless steel in boiling magnesium chloride solution.

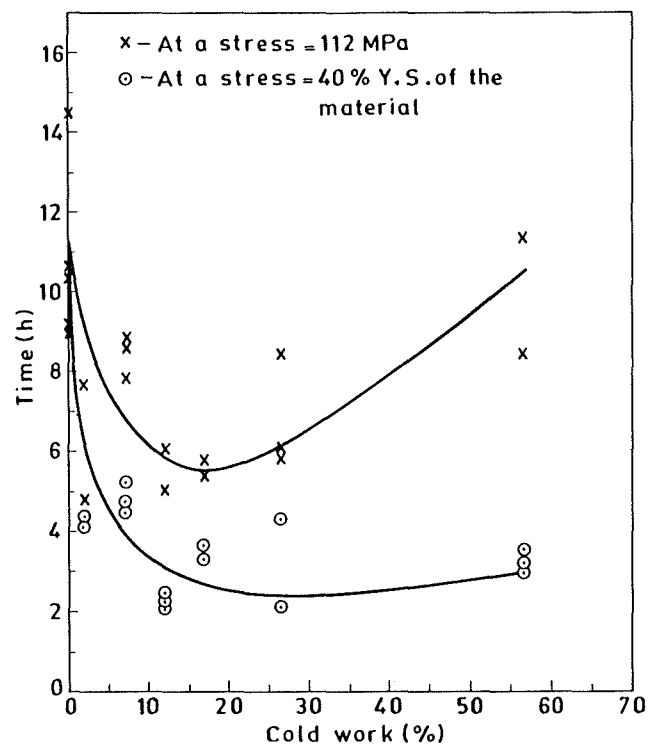
Type 304 Stainless Steel: Stress Elongation Curves



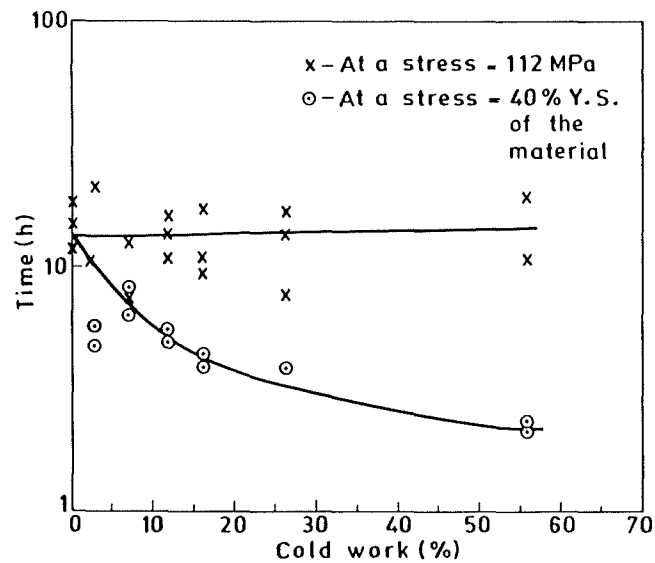
Stress elongation curves for type 304 weld metal specimens tested in boiling magnesium chloride at 154 °C.

Type 316 Stainless Steel: Effect of Cold Work on Time-to-Fracture

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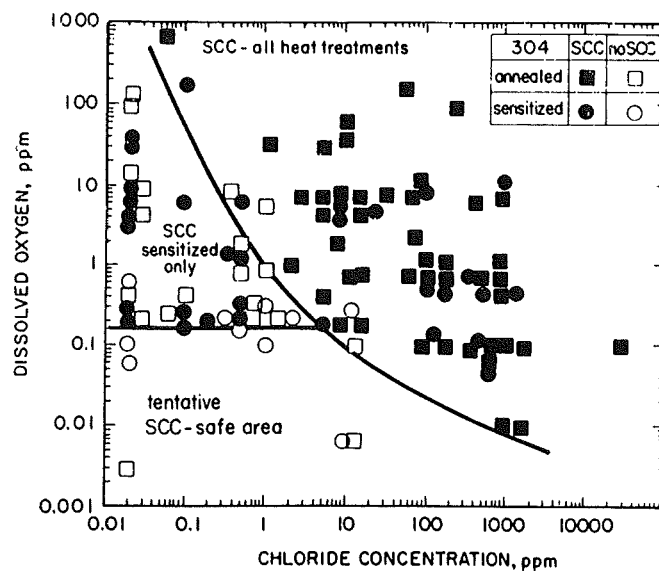
Effect of cold work on the time-to-fracture for type 316 stainless steel.



Effect of cold work on the time-to-fracture for type 304 stainless steel.

Source: P. Muraleedharan, et al., Effect of Cold Work on Stress Corrosion Cracking Behavior of Types 304 and 316 Stainless Steels, Met. Trans. A, 16A, February 1985, p 285-289.

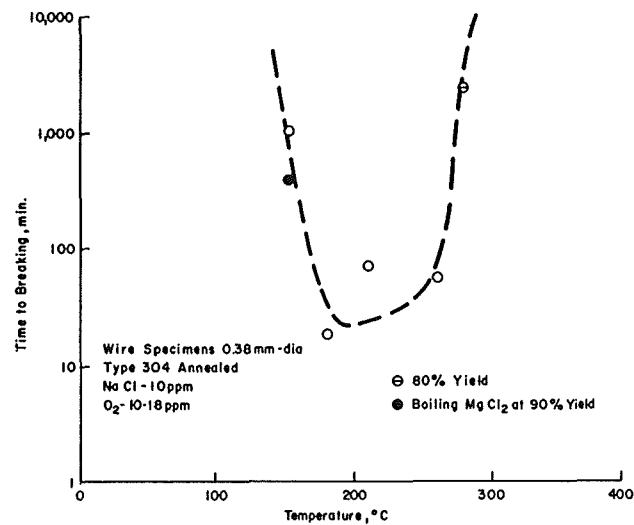
Type 304 Stainless Steel: Effect of Dissolved Oxygen and Chloride on Stress-Corrosion Cracking



Concentration ranges of dissolved oxygen and chloride that may lead to stress-corrosion cracking of type 304 stainless steel in water at 260 to 300 °C. Applied stresses in excess of yield strength and test times in excess of 100 h, or strain rates greater than $10^{-5}/s$.

Type 304 Stainless Steel: Time-to-Breaking at 100% of Yield Stress

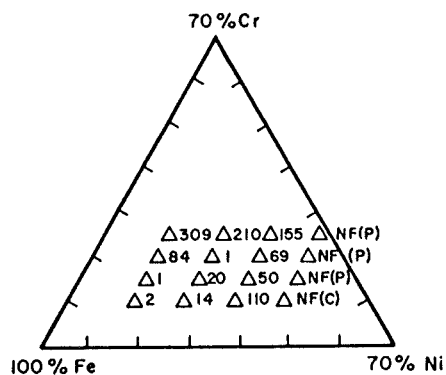
175



Time-to-breaking of first specimen at 100% of yield stress for type 304 stainless steel versus temperature.

Source: J.W. Frey and R.W. Staehle, Effect of Temperature, Stress and Alloy Composition on the Role of Stress Corrosion Cracking in Fe-Ni-Cr Alloys, in High Purity Water Corrosion of Metals, National Association of Corrosion Engineers, Houston, 1968.

Austenitic Steel Ternary Alloys: Stress-Corrosion Cracking Data

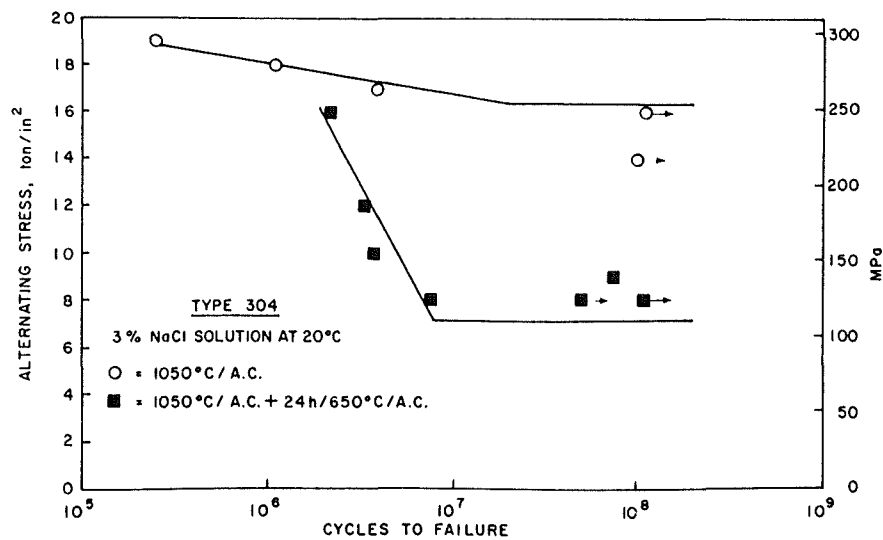


Stress-corrosion cracking data for austenitic steel ternary alloys.

Specimens were loaded at 140 MN/m^2 in 3% sodium chloride solution, under a pressure of 14 MPa oxygen. Numbers are times-to-failure (hours); NF(P) = not failed, but pitted, NF(c) = not failed, but cracked.

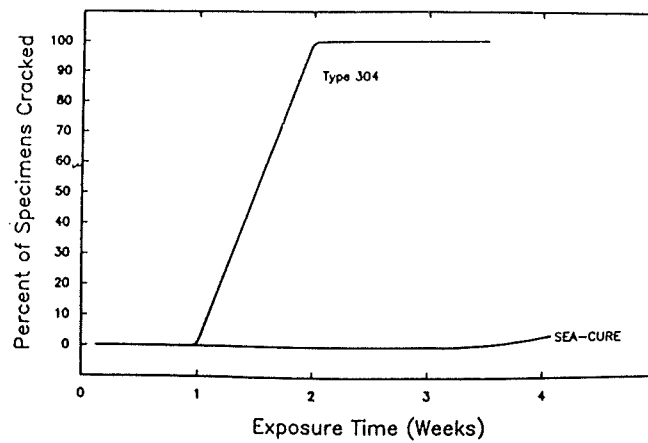
Type 304 Stainless Steel: Effect of Heat Treatment on Corrosion Fatigue, in NaCl

177



Effect of tempering temperature on the corrosion fatigue behavior of type 420 stainless steel.

Type 304 Stainless Steel: Susceptibility to Stress-Corrosion in Chloride Water

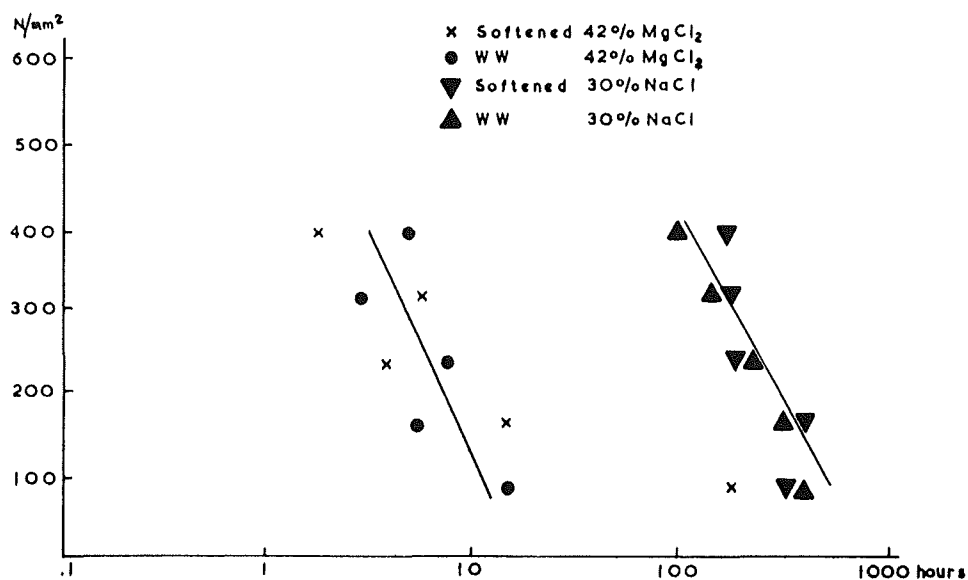


Comparison of susceptibility of austenitic type 304 stainless steel to that of ferritic S44660 stainless (Sea-Cure) to stress-corrosion in 450 °F 100 ppm chloride water.

The alloys were tested as U-bend specimens in oxygen-saturated pH 9 water, ASTM Sea Salt. Under the test conditions, one of 32 Sea-Cure test specimens developed a stress-corrosion crack after 4 weeks of testing, whereas all 32 type 304 specimens failed within 2 weeks.

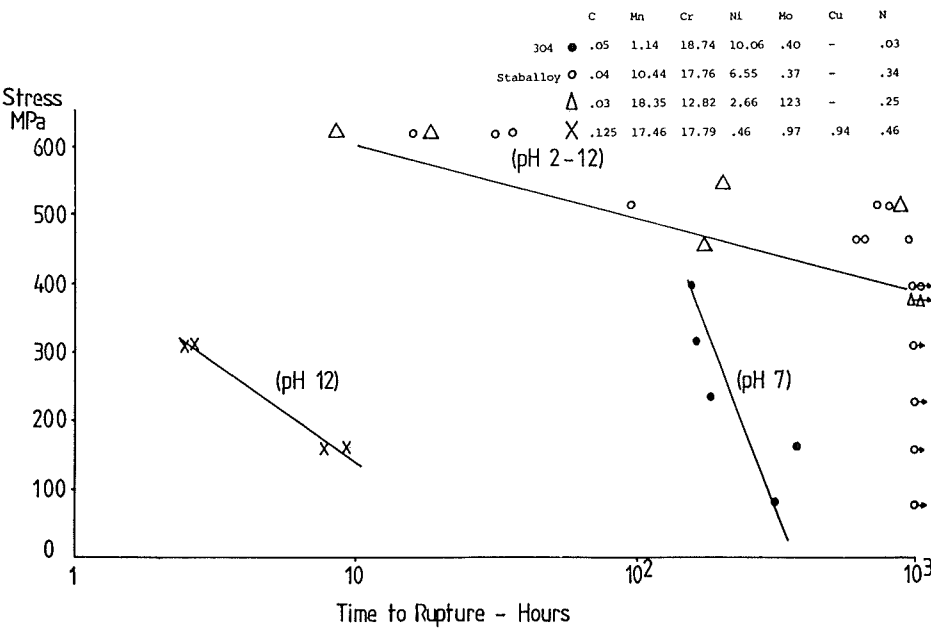
Type 304 Stainless Steel: Time-to-Rupture in Boiling Chloride Solutions

179



Time-to-rupture of type 304 stainless in boiling solutions of magnesium chloride or sodium chloride. The type 304 was tested in a warm worked condition and a softened condition.

Type 304 Stainless Steel: Test Results in Boiling Saturated Sodium Chloride Solutions at Various Levels

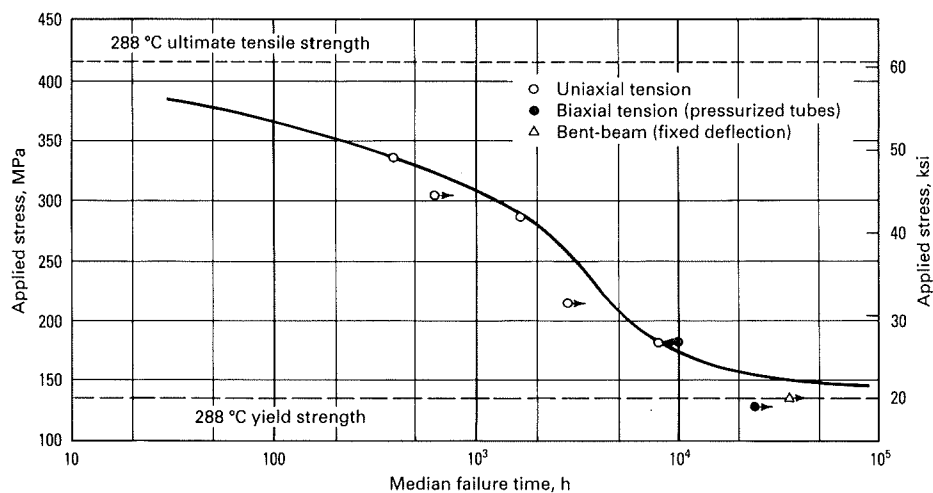


Stress-corrosion cracking test results for type 304 and Staballoy stainless steels obtained in boiling saturated sodium chloride solutions at various pH levels.

Source: J.E. Truman and K.B. Lomax, Stress Corrosion Cracking of Steel Components Used in Directional Drilling of Oil and Gas Wells, in Stainless Steels '87, The Institute of Metals, London, 1988, p 273-283.

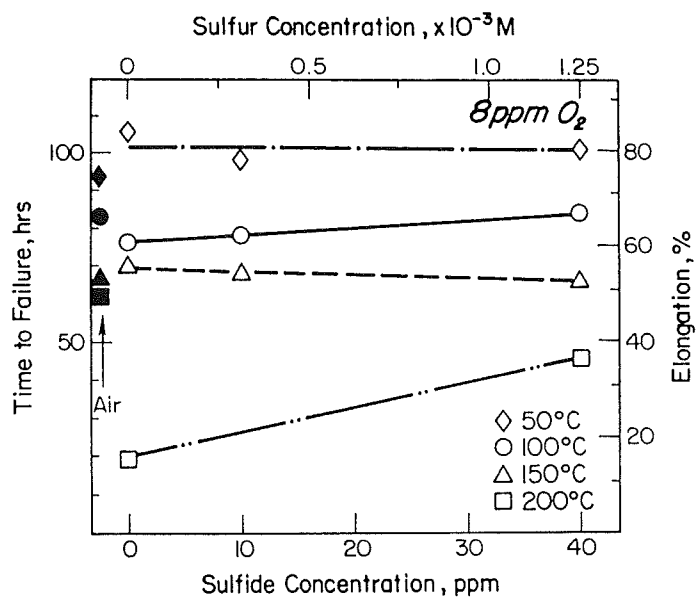
Type 304 Stainless Steel: Stress Dependence of Intergranular Stress-Corrosion Cracking

181



Stress dependence of intergranular stress-corrosion cracking of sensitized type 304 stainless steel in 288 °C (550 °F) water with 0.2 mg/L (ppm) oxygen.

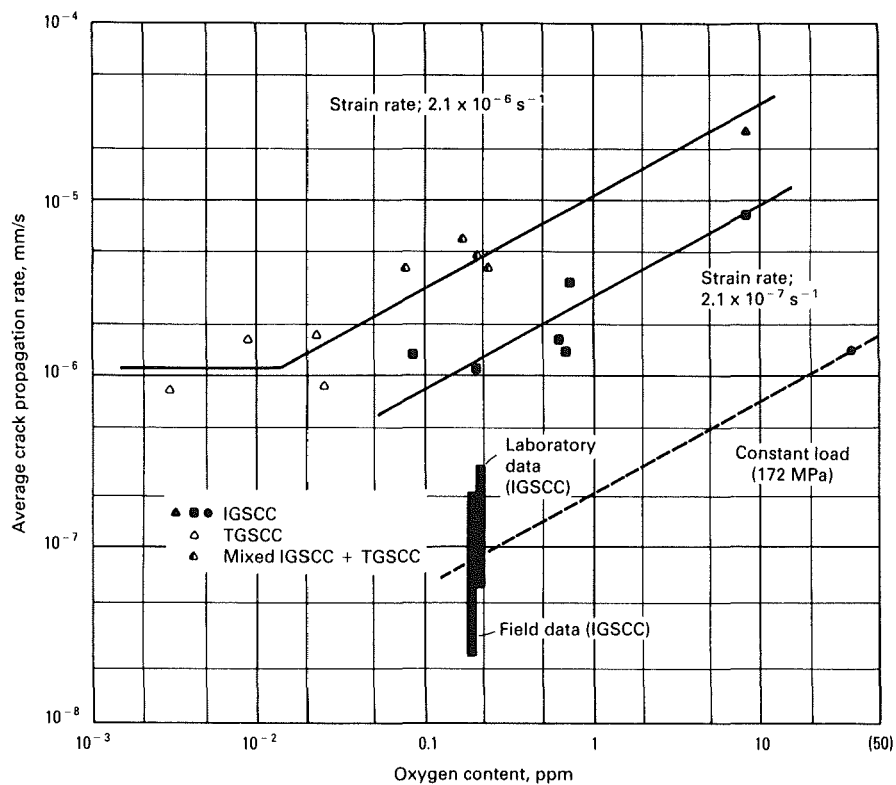
Type 304 Stainless Steel: Effect of Sulfide Ion and Sulfur Concentration on Time-to-Failure



Effect of sulfide ion and sulfur concentration on the time-to-failure and elongation-to-fracture of specimens tested in aerated water at various temperatures.

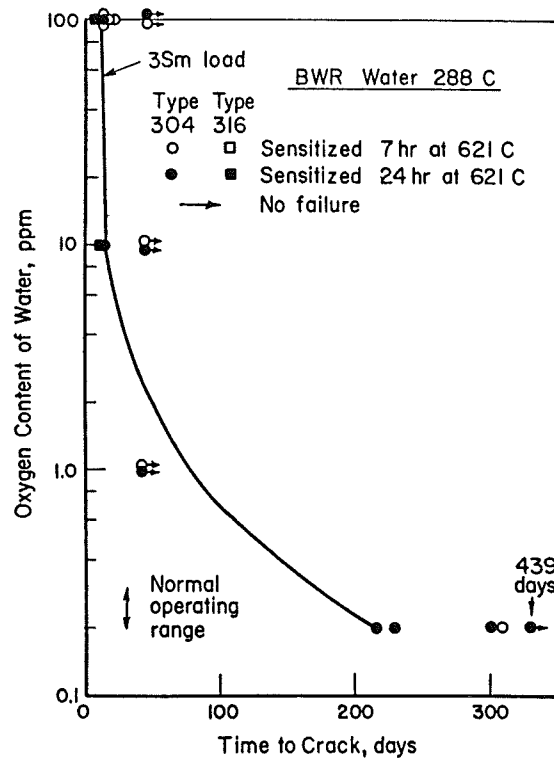
Type 304 Stainless Steel: Variation in Average Crack Propagation Rate

183



Variation in the average crack propagation rate in sensitized type 304 stainless steel in water at 288 °C (550 °F) with oxygen content. Data are from constant extension rate tests, constant load, and field observations on boiling water reactor piping.

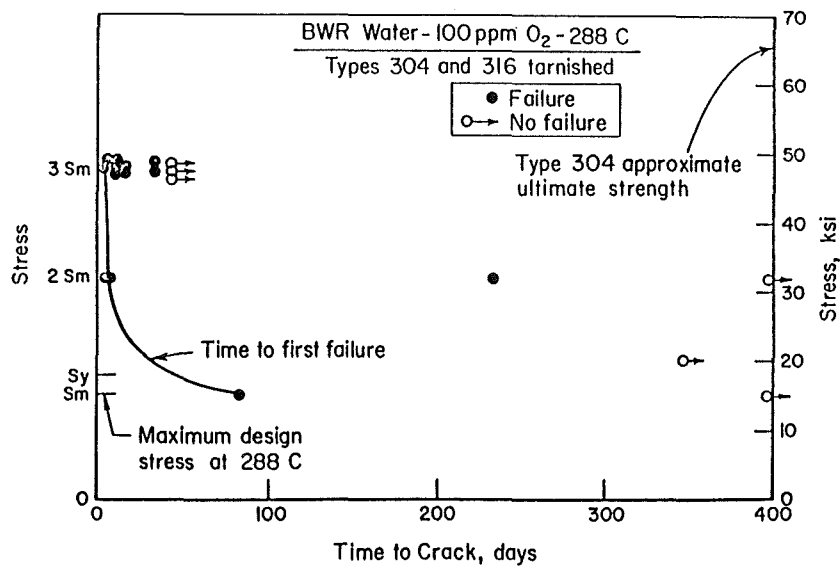
Sensitized Stainless Steel: Effect of Dissolved Oxygen on Time-to-Cracking



Effect of dissolved oxygen level on time-to-cracking of sensitized stainless steel specimens stressed to 3 Sm in water at 288 °C (Sm = 90% of yield stress at 288 °C).

Sensitized Stainless Steel: Effect of Stress Level on Time-to-Cracking

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Effect of stress level on the time-to-cracking of sensitized stainless steel exposed to 288 °C water containing 100 ppm dissolved oxygen.

Type 304 Stainless Steel: Intergranular Stress-Corrosion Cracking

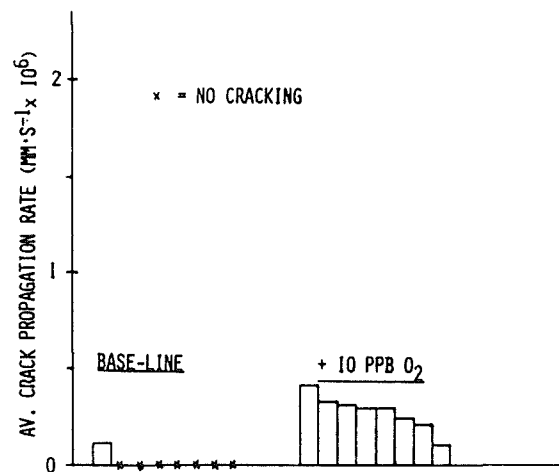
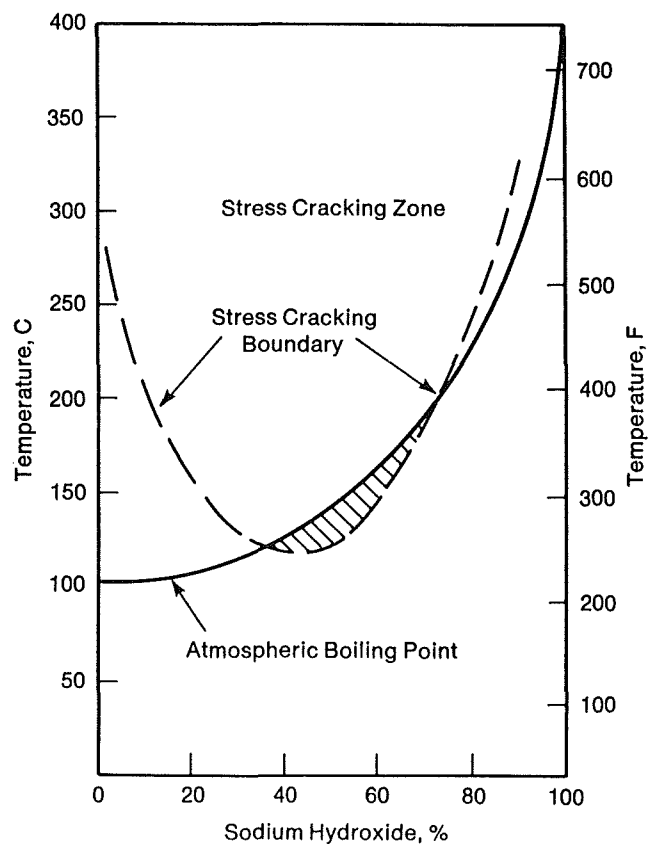


Figure 2. CERT test results on IGSCC in Type 304 SS from Test A. Approximate base-line environment composition 4 ppb O₂, 140 ppb H₂

CERT test results on intergranular stress-corrosion cracking in type 304 stainless steel in BWR environment containing impurities. The baseline BWR environment contained approximately 4 ppb O₂ and 140 ppb H₂.

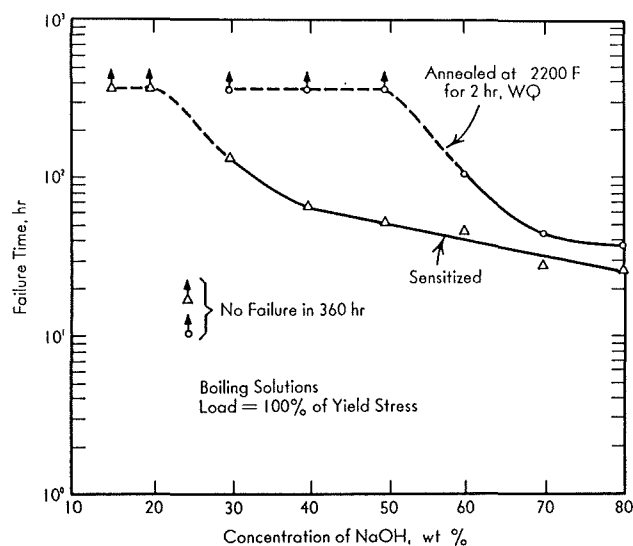
Type 304 Stainless Steel: Stress-Corrosion Cracking in Sodium Hydroxide

187



Stress-corrosion cracking behavior of type 304 stainless steel in sodium hydroxide.

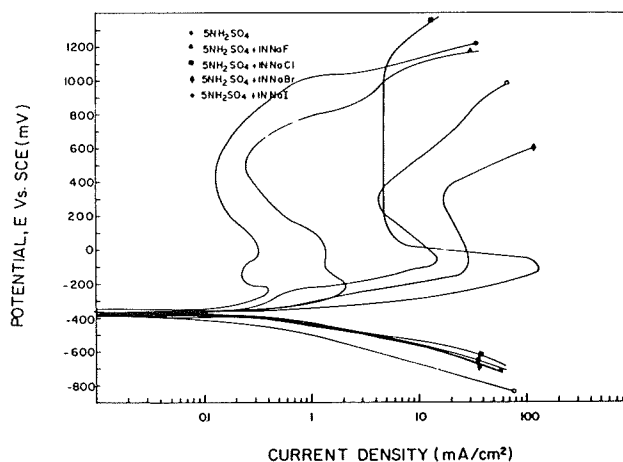
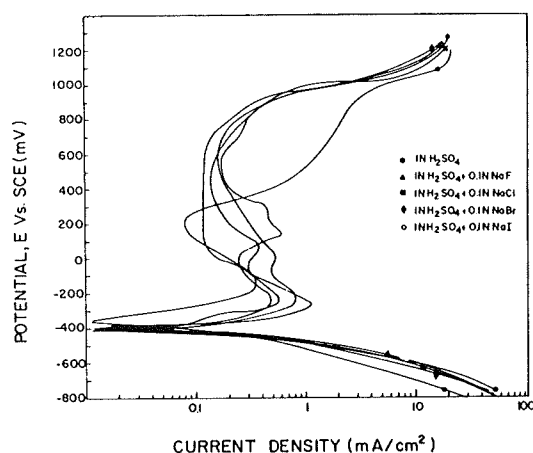
Type 304 Stainless Steel: Stress-Corrosion Cracking in Caustic Soda Solutions



Stress-corrosion cracking of annealed and sensitized type 304 stainless steel in caustic soda solutions.

Type 304 Stainless Steel: Potentiodynamic Polarization Curves

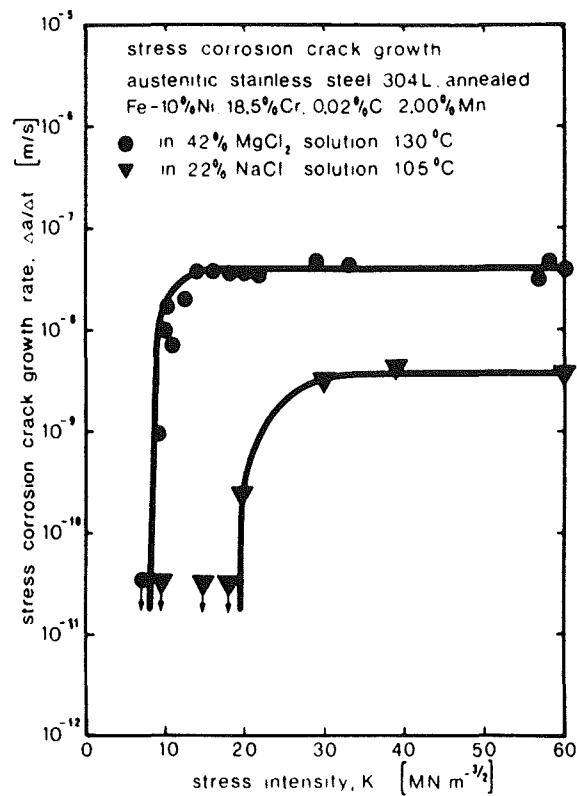
189



Potentiodynamic polarization curves of type 304 stainless steel tested in the indicated solutions.

Stress-corrosion cracking of U-bend specimens of type 304 stainless was not observed in 1N H₂SO₄, but was observed in 5N H₂SO₄ + 1N NaBr solutions. Type 304 is in the active state in 5N H₂SO₄, and in the passive state in 1N H₂SO₄. Addition of alkali halide salts increased the corrosion rates. It was suggested that the main crack formed as the result of anodic dissolution, which caused the propagation of corrosion tunnels.

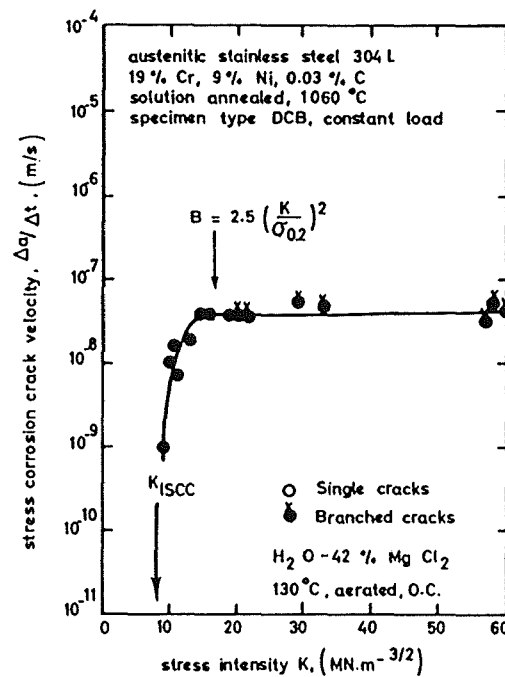
Type 304L Stainless Steel: Effect of Stress-Intensity on Growth Rate of Stress-Corrosion Cracks



Effect of stress-intensity on the growth rate of stress-corrosion cracks in austenitic type 304L stainless steel. Higher temperatures and higher chloride concentrations result in higher crack velocities and a lower threshold stress-intensity.

Type 304L Austenitic Stainless Steel: Crack Propagation as a Function of Stress-Intensity in Precracked Samples

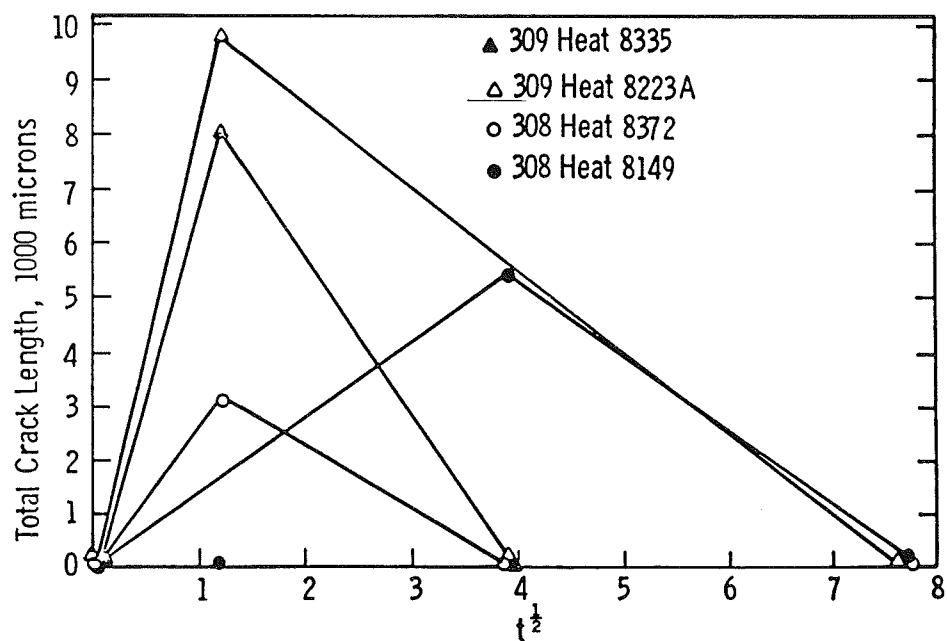
191



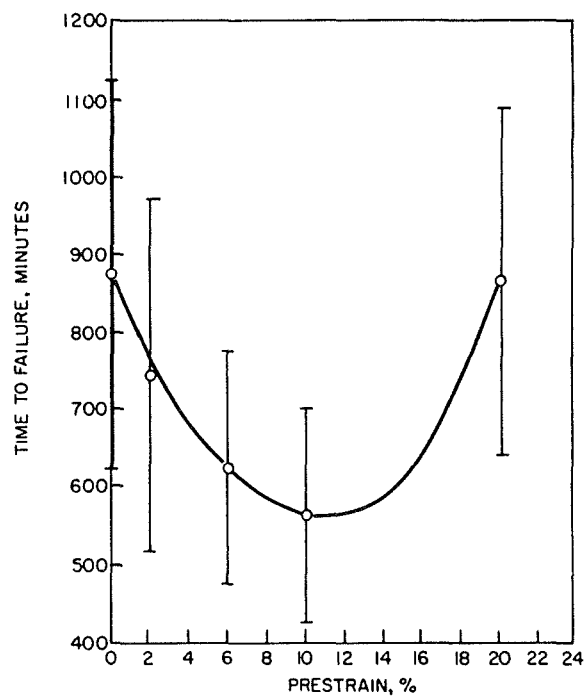
Crack propagation as a function of stress-intensity in precracked samples.

Plane-strain conditions depend upon the thickness, B, and hold for stress-intensity levels below the indicated value. Above this value, plane-stress conditions increase with an increase in stress-intensity.

Type 308 and 309 Stainless Steels: Impact of Postweld Heat Treatment on Stress-Corrosion Susceptibility

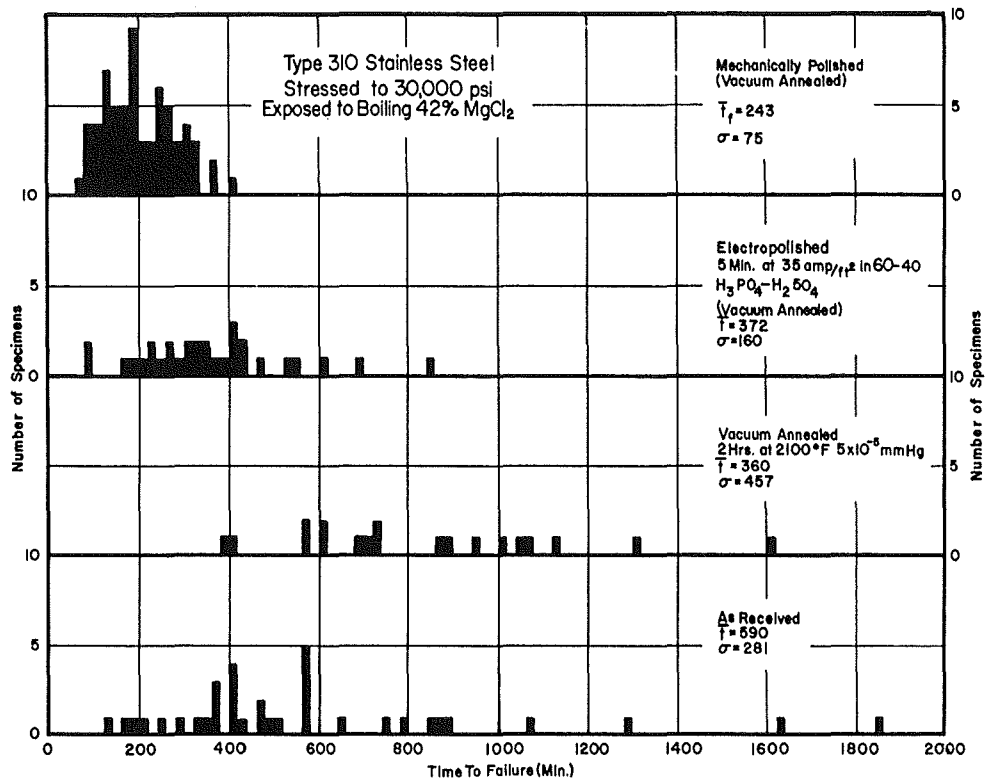


Variation of stress-corrosion susceptibility with duration of postweld heat treatment.



Effect of prestrain on time-to-failure of type 310 stainless steel exposed to a magnesium chloride solution boiling at 154 °C and stressed at 90% of the yield stress.

Type 310 Stainless Steel: Effect of Surface Preparation on Time-to-Breaking

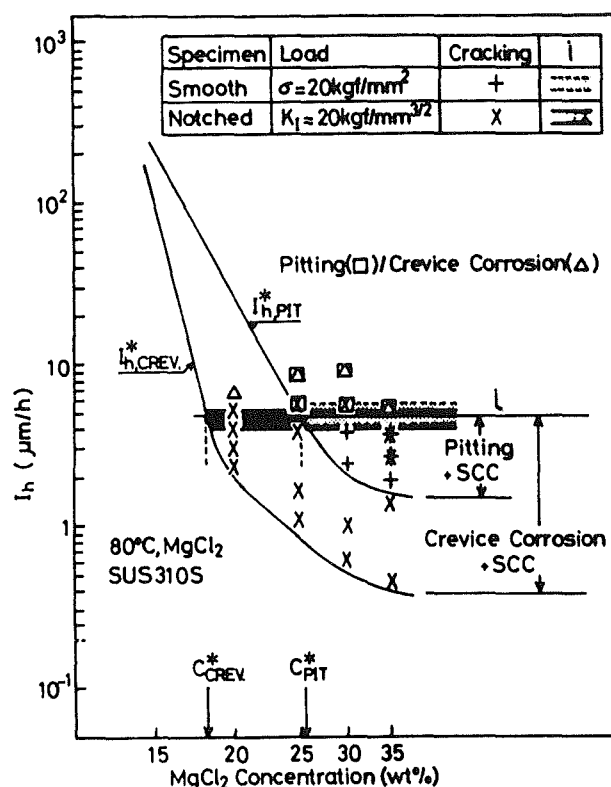


Effect of surface preparation on time-to-breaking of type 310 stainless steel.

Source: R.W. Staehle, Circulating Autoclave System for Stress Corrosion Cracking Studies, in Stress Corrosion Testing, STP 425, American Society for Testing and Materials, Philadelphia, 1967, p 248-273.

Type 310S Steel: Stress-Corrosion Cracking Regions for Smooth and Spark-Machined Notched Specimens

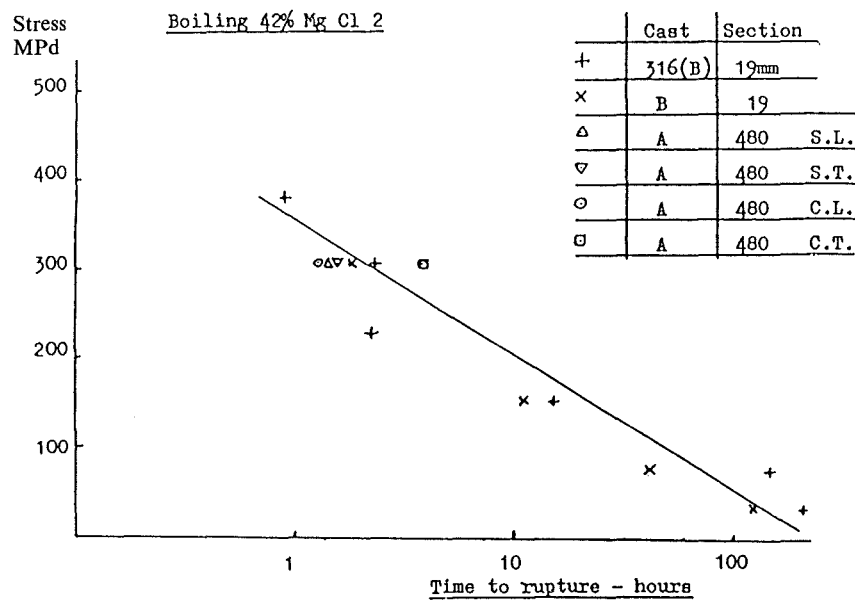
195



Comparison of the stress-corrosion cracking region for smooth specimens and spark-machined notched specimens in terms of the dissolution rate, I_h , and magnesium chloride concentration at 80 °C.

The crack growth rate, i , which was regarded as a plateau velocity, did not depend on magnesium chloride concentration, electrode potential, or crack initiation site. Cracks from pits in smooth specimens occurred in the region $I_{h,pit}^* < I_h < i$. Cracks in crevices in notched specimens occurred in the region $I_{h,crev}^* < I_h < i$. The intersection of $I_{h,i}^*$ ($i = \text{pit, crev}$) with the crack growth rate, i , gives the critical concentration of magnesium chloride (C_i^*), below which there exists no stress-corrosion cracking region. In higher concentrations of magnesium chloride, $I_{h,i}^*$ became smaller and the transition to stress-corrosion cracking occurred more readily.

Austenitic Stainless Steel: Initial Stress Versus Time-to-Rupture in Boiling Magnesium Chloride Solution

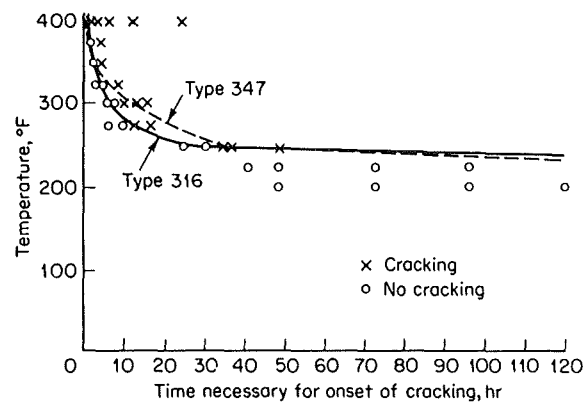


Initial stress versus time-to-rupture for samples tested in boiling 42% magnesium chloride solution.

Source: J.E. Truman and K.B. Lomax, Effect of Section Size on the Properties of a Nitrogen-Bearing Austenitic Steel of High Corrosion Resistance, in High Nitrogen Steels, The Institute of Metals, London, 1989, p 414-422.

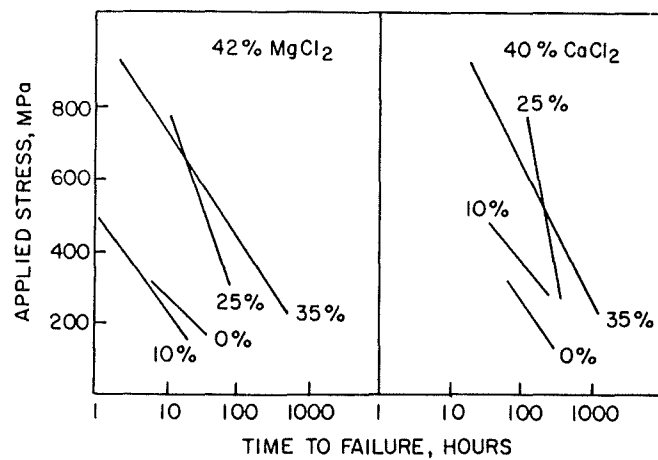
Types 316 and 347 Stainless Steel: Effect of Temperature on Initiation of Stress-Corrosion Cracking

197



Effect of temperature on the time for the initiation of stress-corrosion cracking in types 316 and 347 stainless steels in water containing 875 ppm sodium chloride.

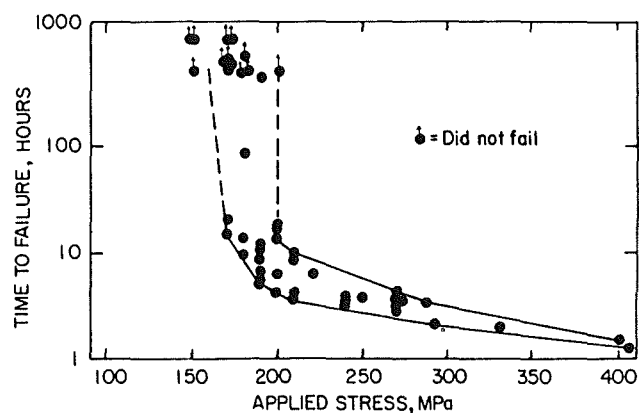
Type 321 Stainless Steel: Effect of Cold Work on Susceptibility to Cracking



Effect of cold work (%) on the susceptibility to cracking of type 321 stainless steel in boiling magnesium chloride and in calcium chloride solutions.

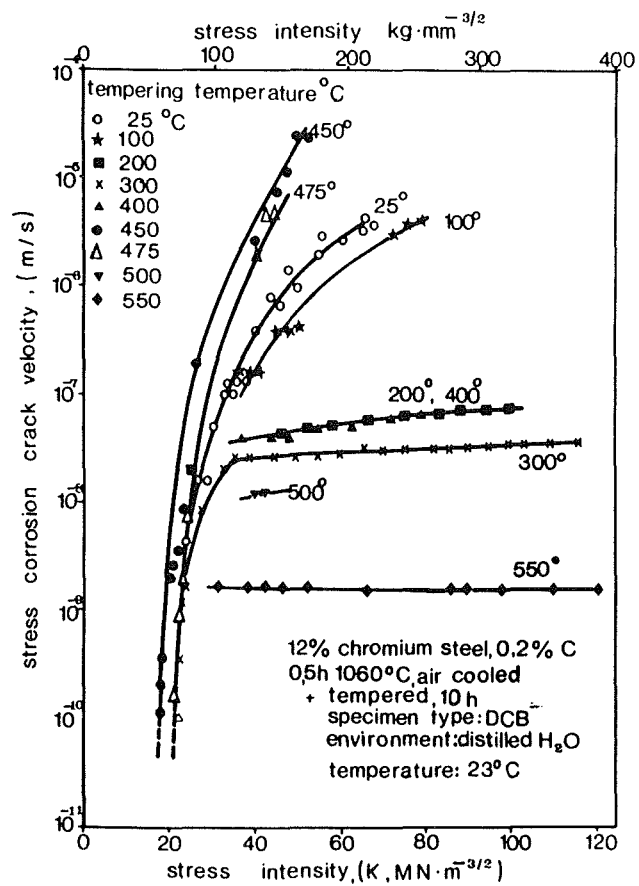
Type 347 Stainless Steel: Effect of Applied Stress on Time-to-Failure

199



Effect of applied stress on the time-to-failure of solution annealed and electropolished type 347 stainless steel tested in boiling magnesium chloride at 145 °C.

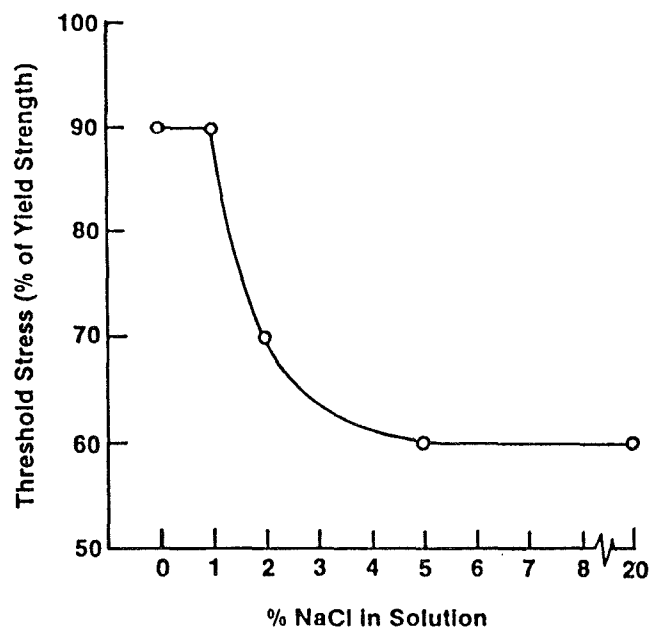
12Cr Steel: Impact of Tempering Temperature and Stress-Intensity on Stress-Corrosion Cracks



Effect of tempering temperature and stress-intensity on the velocity of stress-corrosion cracks in a 12Cr steel.

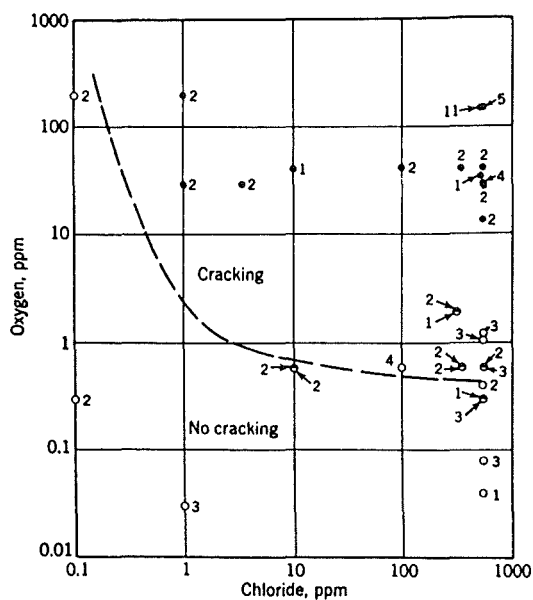
13Cr Stainless Steel: Effect of Sodium Chloride Concentration in Solution on Threshold Stress

201

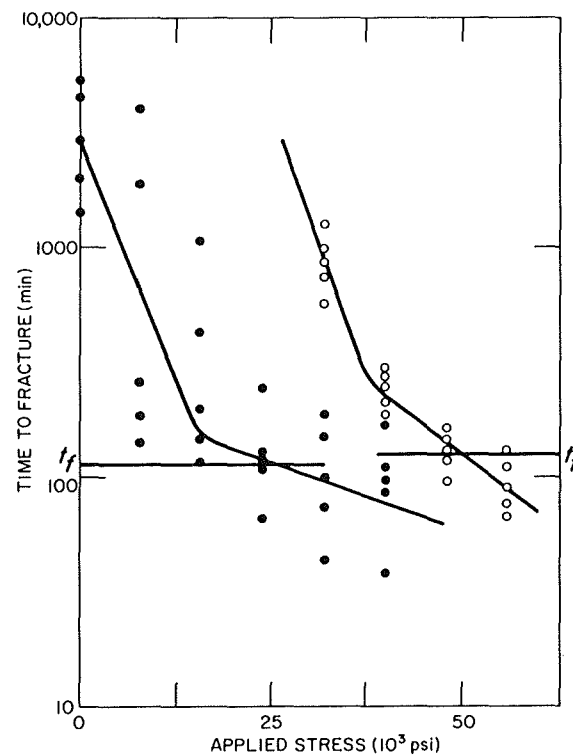


Effect of sodium chloride concentration in NACE solution on the threshold stress for 620 MPa (90 ksi) yield strength in 13Cr stainless steel.

18-8 Stainless Steel: Susceptibility to Stress-Corrosion Cracking Failure as a Function of Chloride and Oxygen Environments

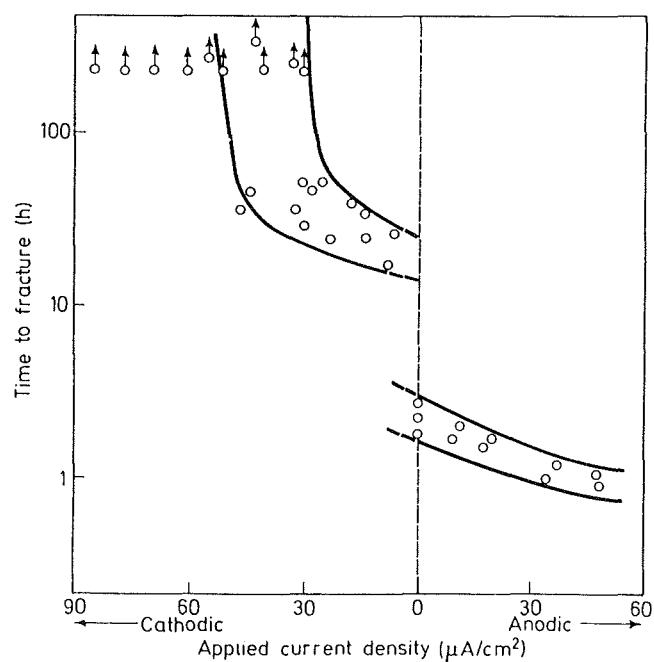


Susceptibility of 18-8 type austenitic stainless steel to stress-corrosion cracking failure as a function of the oxygen and chloride content of the environment (steam and intermittent wetting, 515 to 535 K, pH 10.6, 1 to 30 days exposure). Numbers inside figure refer to the number of specimens.

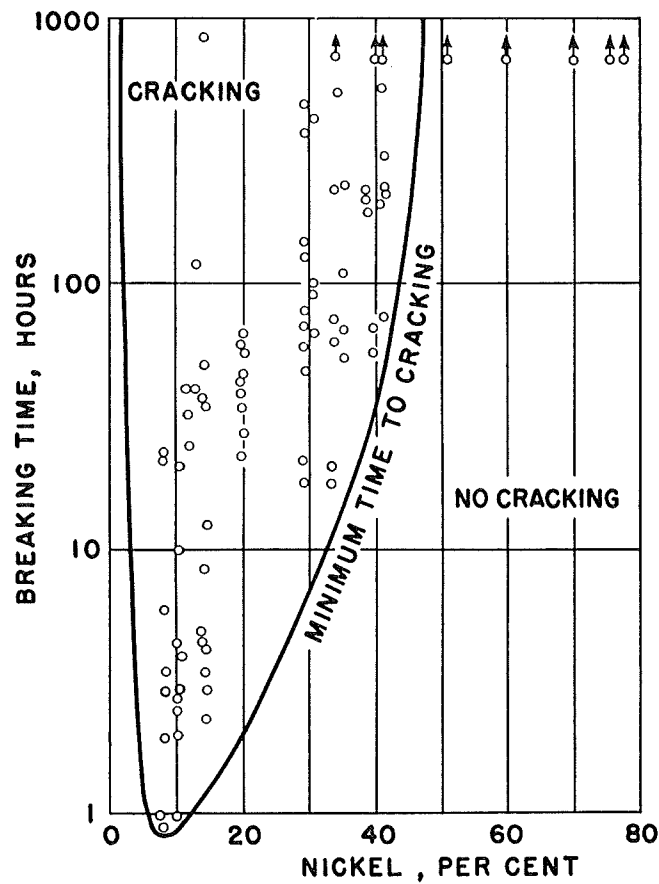


Relation of applied stress to time-to-fracture of austenitic stainless steels exposed to magnesium chloride at 152 to 153 °C. Open symbols: 18-8; closed symbols: 18-10-3MoTi.

Fe-18Cr-8Ni Steel: Effect of Anodic or Cathodic Polarization on Time-to-Fracture



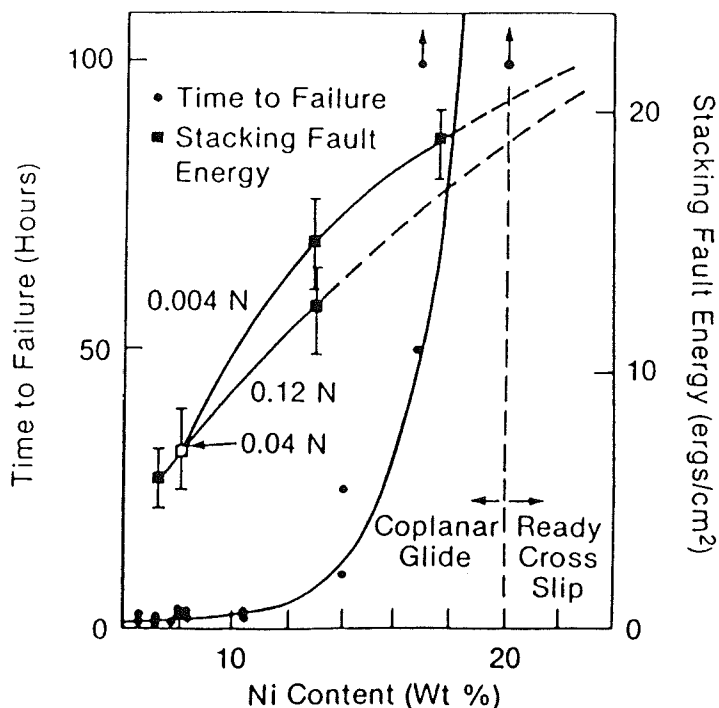
Effect of anodic or cathodic polarization on time-to-fracture of Fe-18Cr-8Ni steel.



Stress-corrosion cracking characteristics of Fe-18Cr-Ni wires in boiling 42% magnesium chloride.

The wires were stressed at 90% of yield strength. Alloys containing nickel in excess of 40% appear completely resistant to chloride stress-corrosion cracking.

Fe-18Cr Alloys: Effect of Nickel Content on Time-to-Failure

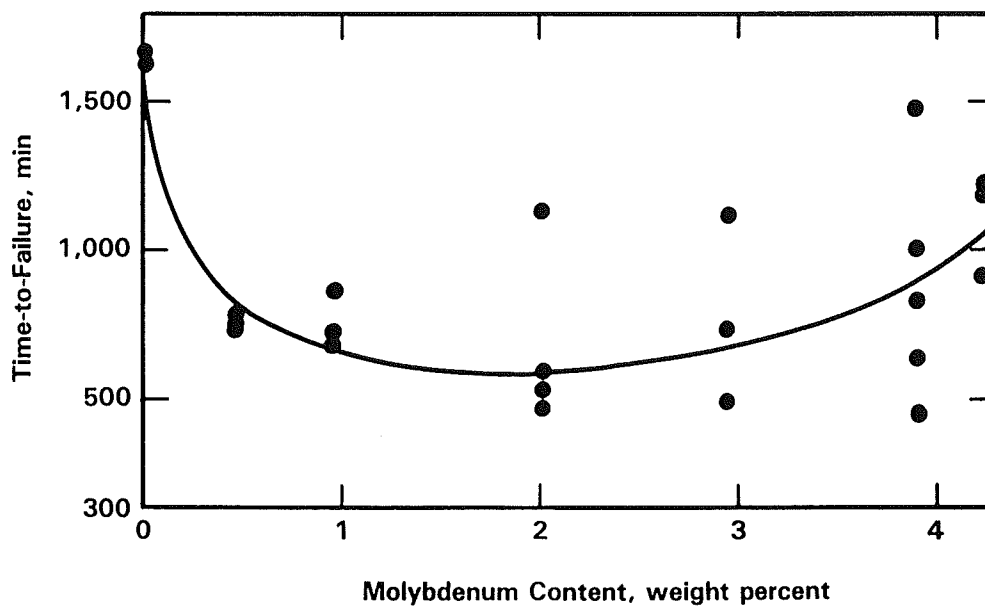


Effect of nickel content on time-to-failure of Fe-18Cr alloys in boiling aqueous magnesium chloride.

The mode of cracking was transgranular. Because nickel increases the resistance to stress-corrosion cracking and also raises the stacking fault energy, it has been proposed that this change in stacking fault energy is related to the increase in resistance. An increase in stacking fault energy would promote cross-slip and make the slip process less planar, thereby reducing stress concentrations at the tips of blocked slip bands which otherwise would promote the rupture of the protective surface oxide film. A low stacking fault is not an essential for cracking, because cracking has been observed in stainless steels in which the dislocations were not arranged in planar arrays.

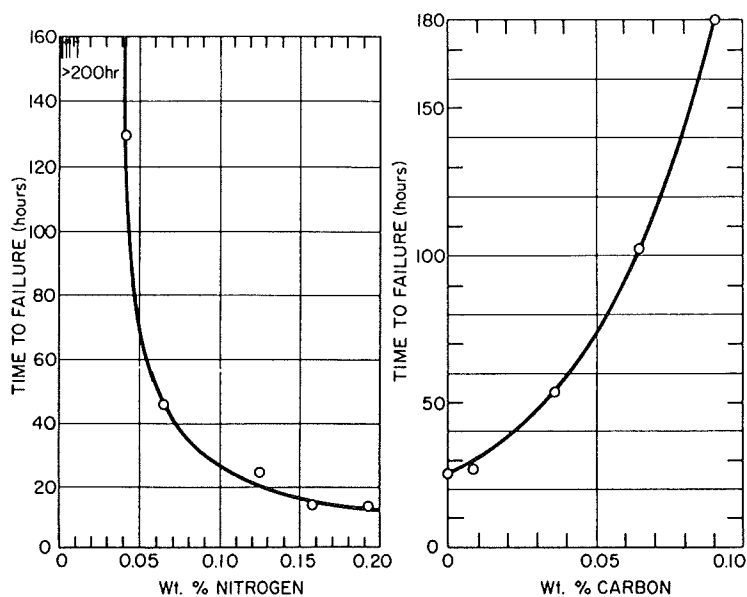
**18Cr-14Ni-0.4.2Mo Steel: Time-to-Failure in Boiling
Magnesium Chloride**

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Time-to-failure results obtained on 18Cr-14Ni-0.4.2Mo steels stressed at 75% of room-temperature yield strength in boiling magnesium chloride (140 °C).

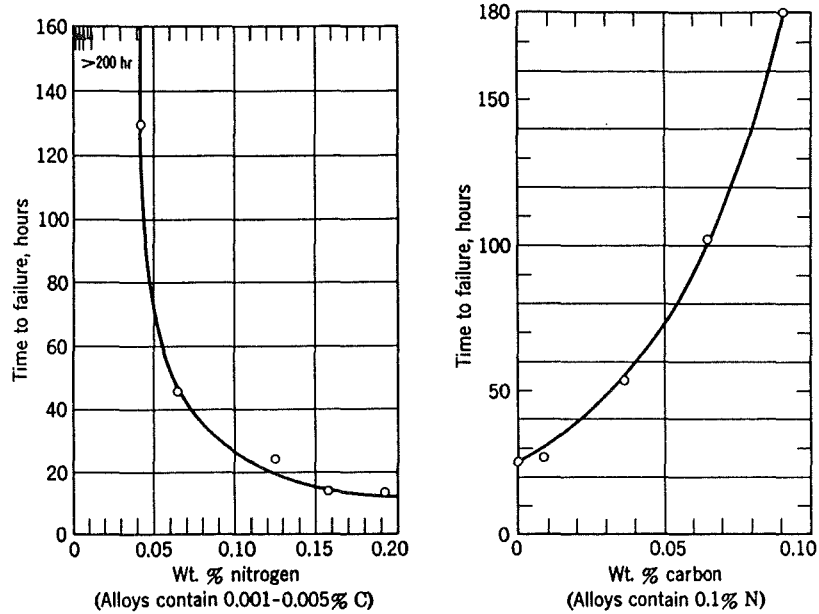
19Cr-20Ni Austenitic Stainless Steel: Stress-Corrosion Cracking in Boiling Magnesium Chloride



Stress-corrosion cracking of cold rolled 19Cr-20Ni austenitic stainless steels in boiling magnesium chloride (154°C) as affected by carbon or nitrogen content.

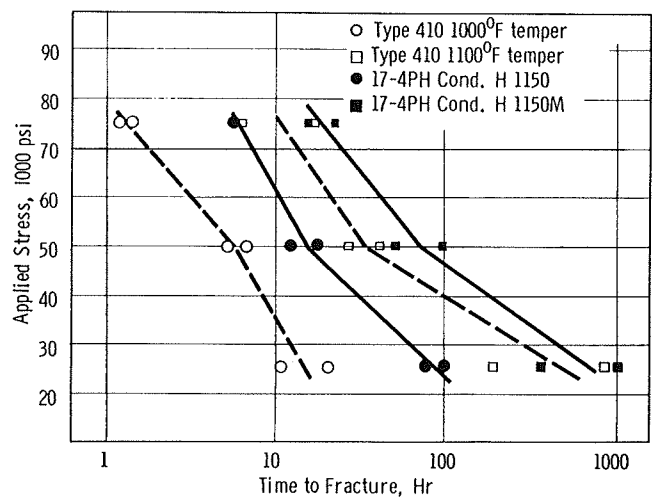
19Cr-20Ni Austenitic Stainless Steel: Stress-Corrosion Cracking as Affected by Carbon or Nitrogen Content

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Stress-corrosion cracking of cold rolled 19Cr-20Ni austenitic stainless steels in boiling magnesium chloride (154°C) as affected by carbon or nitrogen content.

Types 410 and 17-4 pH Steels: Stress-Corrosion Test Results

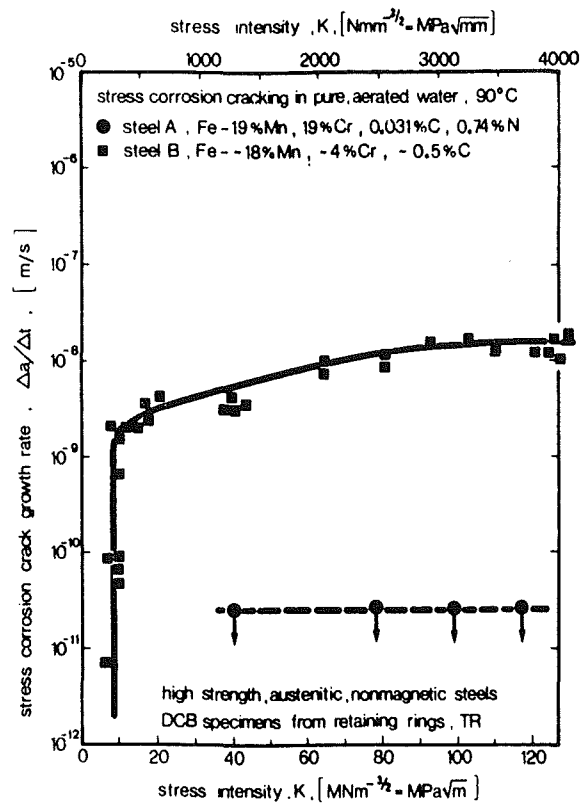


Stress-corrosion test results in sodium chloride/acetic acid aqueous solution saturated with hydrogen sulfide for types 410 and 17-4 pH steels.

Stress corrosion resistance of Type 410 and 17-4 PH steels.

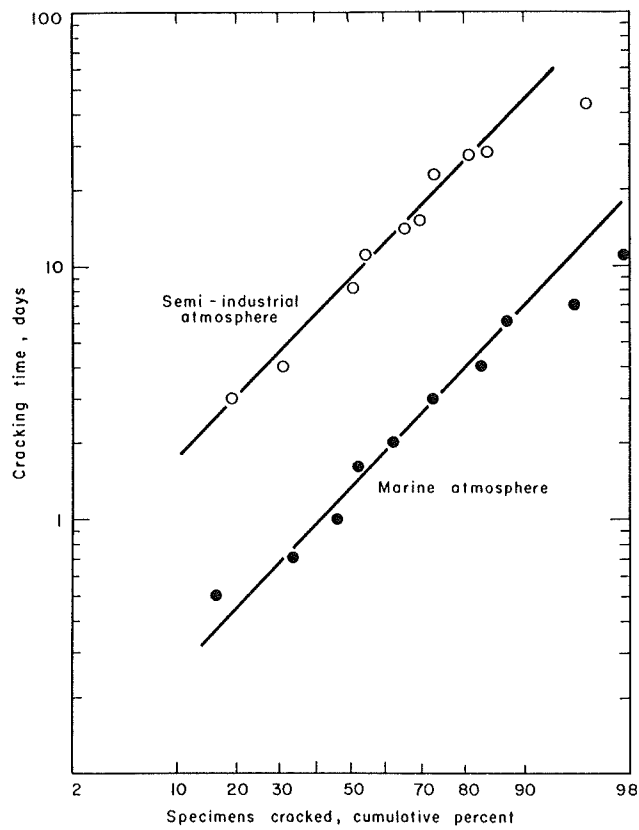
Alloy	Heat Treatment	0.2% Yield Strength, psi	Applied Stress, psi	Time-to-Failure, hr ^a
Type 410. . . .	1800 F, ½ hr, oil quench + 1000 F, 4 hr, air cool	125 000	75 000	1.3, 1.4
			50 000	5.7, 6.5
			25 000	11, 20
Type 410. . . .	1800 F, ½ hr, oil quench + 1100 F, 4 hr, air cool	100 000	75 000	6.2, 18
			50 000	27, 39
			25 000	195, 845
17-4 PH. . . .	1925 F, ½ hr, air cool + 1150 F, 4 hr, air cool	125 000	75 000	6.1, 6.1
			50 000	12, 18
			25 000	83, 102
17-4 PH. . . .	1925 F, ½ hr, air cool + 1500 F, 2 hr, air cool + 1150 F, 4 hr, air cool	100 000	75 000	16, 22
			50 000	50, 95
			25 000	345, 1105

^a Tested at room temperature in 6% NaCl-0.5% acetic acid solution, saturated with H₂S.



Stress-corrosion crack velocities of two different retaining ring steels in water. Note that the nitrogen-containing steel is resistant to stress-corrosion cracking in water.

USS 12MoV Stainless Steel: Distribution of Stress-Corrosion Results

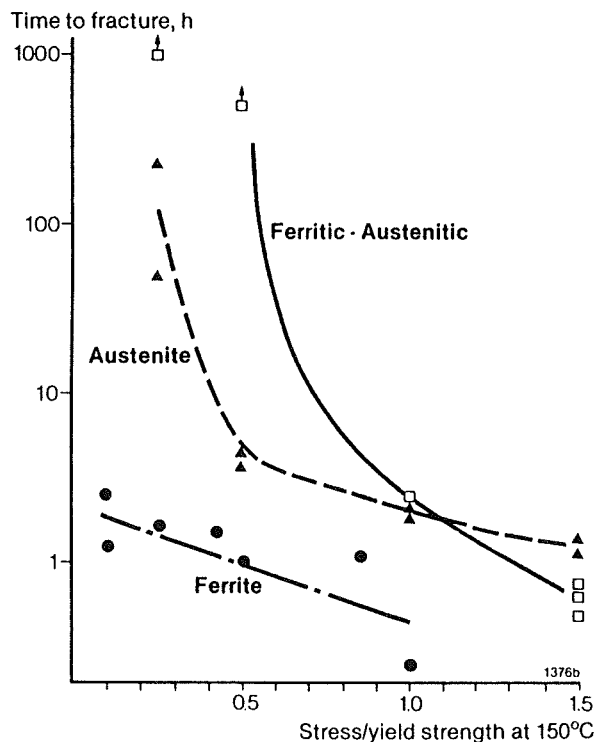


Distribution of stress-corrosion results for USS 12MoV stainless steel.

Source: V.S. Goel, Ed., Corrosion Cracking, American Society for Metals, Metals Park, 1986.

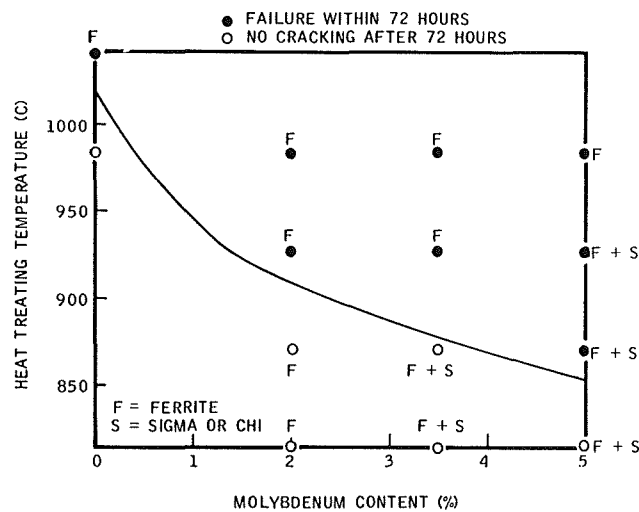
21Cr-8Ni-2Mo-5Cu Duplex Stainless Steel: Comparison of Stress-Corrosion Cracking Behavior of Austenitic and Ferritic Phases, in Magnesium Chloride

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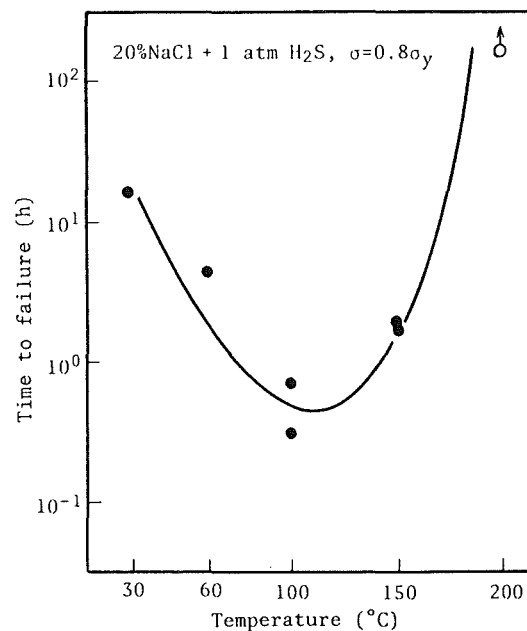
Performance of the duplex alloy 21Cr-8Ni-2Mo-5Cu compared to the performance of alloys whose compositions correspond to its austenite and ferrite phases. Stress-corrosion cracking tests were carried out at constant load in 44% magnesium chloride at 153 °C.

25Cr Ferritic Stainless Steel: Cracking Threshold as a Function of Heat Treating Temperature



Cracking threshold of hydrogen-charged U-bend specimens of high-purity 25Cr ferritic stainless steel as a function of heat treating temperature and molybdenum content.

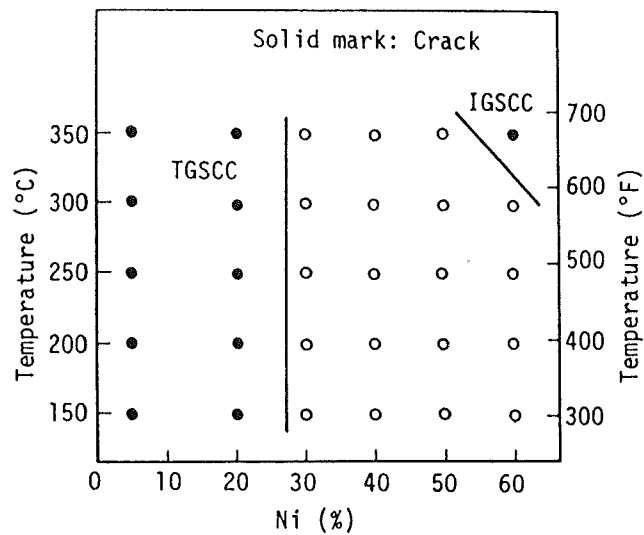
The specimens were exposed at a cathodic current density of 20 mA/cm² in a 5% sulfuric acid solution containing 4 mg of arsenic trioxide per liter of solution.



Stress-corrosion cracking test results for 25Cr duplex stainless steel (30% cold work).

The results are for constant-load tests in 20% sodium chloride solution with 1 atm H₂S. The applied stress was 80% of the yield stress. 25Cr duplex stainless steel has a high resistance to hydrogen embrittlement. However, it is susceptible to hydrogen embrittlement in environments containing H₂S when tested at a stress level in a high-strength condition.

25Cr-3Mo Alloys: Effect of Nickel Content and Solution Temperature on Cracking Behavior

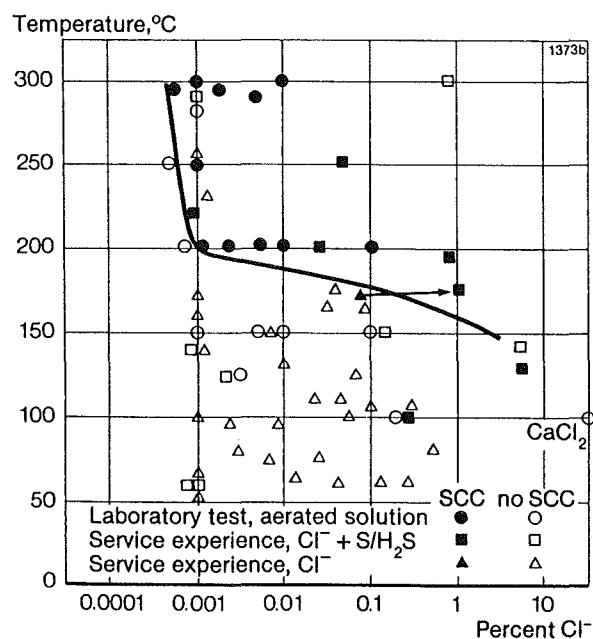


Effect of nickel content and solution temperature on cracking behavior of 25Cr-3Mo alloys in 20% sodium chloride solution with 0.5% acetic acid and 1.0 MPa (150 psi) H₂S and CO₂ partial pressures.

Source: R.D. Kane, Corrosion of Stainless Alloys in Oilfield Production Environments, in Stainless Steels '84, The Institute of Metals, London, 1985, p 429-437.

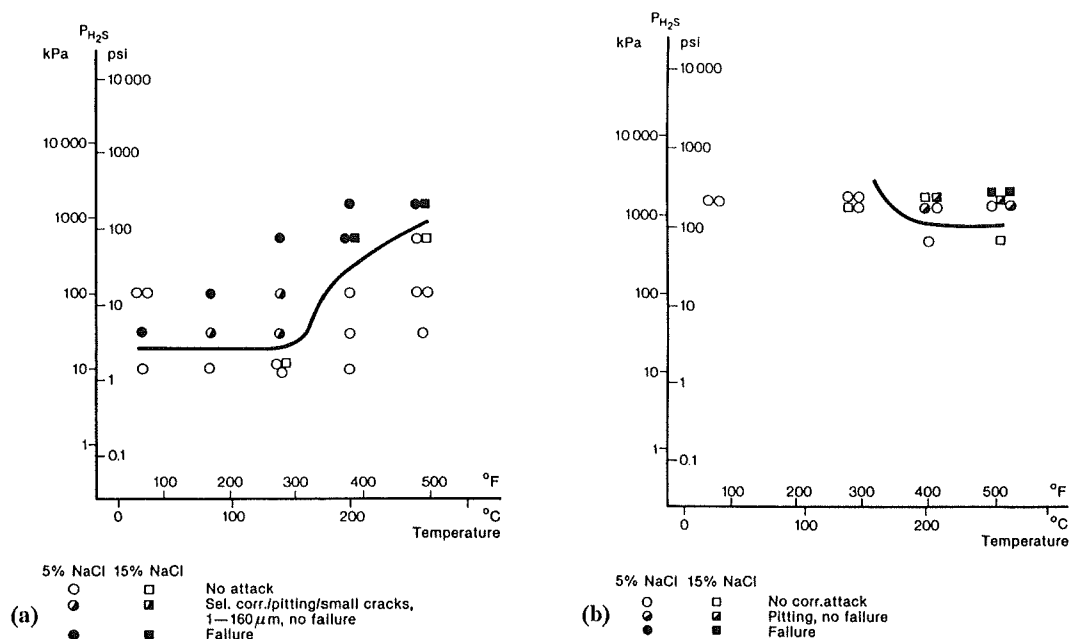
3RE60 Duplex Stainless Steel: Offshore Experience Plus Laboratory Stress-Corrosion Cracking Test Data

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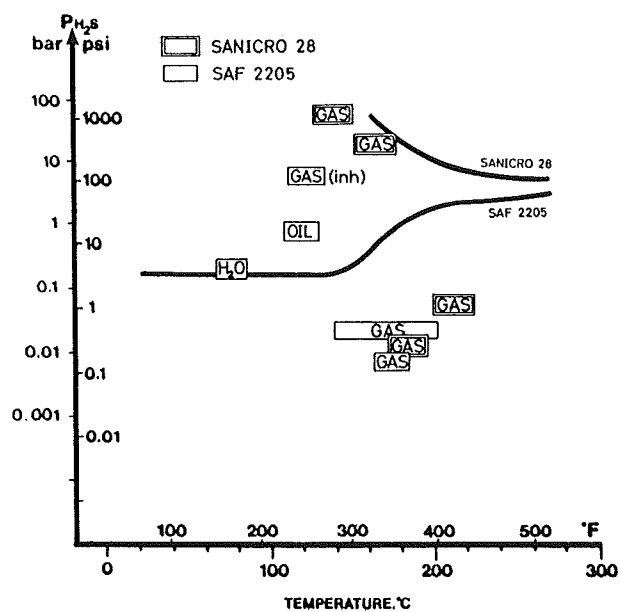
Compilation of practical experience and laboratory stress-corrosion cracking test data for 3RE60 (0.030% C max, 1.7% Si, 1.5% Mn, 0.03% P max, 0.030% S max, 18.4% Cr, 4.9% Ni, 2.7% Mo; microstructure: duplex ferritic-austenitic; 0.2% proof stress, 450 MPa). The practical experience is with topside applications on off-shore platforms.

SAF 2205-130 and Sanicro 28-110 Stainless Steels: Effect of Temperature and Partial Pressure



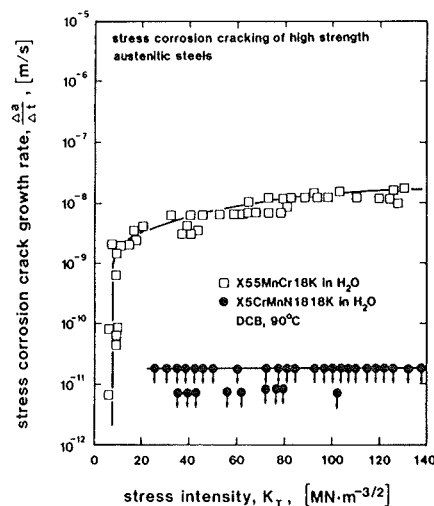
Effect of temperature and hydrogen sulfide partial pressure on behavior of SAF 2205-130 and Sanicro 28-110 stainless steels.

(a) SAF 2205-130 (0.030% C max, 22% Cr, 5.5% Ni, 3.0% Mo, 0.14% N; 0.2% proof stress, 896 MPa). (b) Sanicro 28-110 (0.02% C max, 27% Cr, 3.1% Ni, 3.5% Mo, 1.0% Cu; 0.2% proof stress, 758 MPa). The stress-corrosion cracking tests were carried out in NACE solution for 500 h, at sodium chloride levels of 5 and 15%. The partial pressure of CO₂ was maintained at 70 bar. In discussing oil field corrosion, the term "sweet" corrosion denotes corrosion due to CO₂, and "sour" refers to corrosion due to H₂S. The capability of stainless steels to resist sweet corrosion is known to be good, whereas stainless steels are less resistant to sour corrosion.



Well conditions for some deliveries of SAF 2205 and Sanicro 28. The applicable limits based upon the stress-corrosion cracking tests are also shown.

Austenitic Manganese Steel: Effect of Nitrogen Content on Stress-Corrosion Cracking

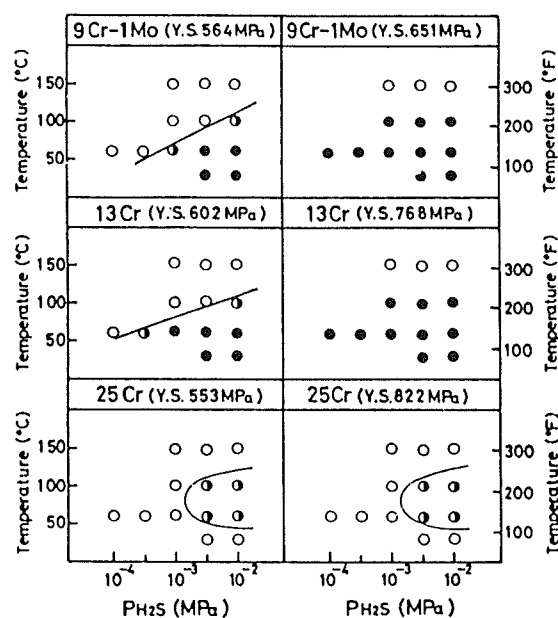


Comparison of the stress-corrosion crack curves of two high-strength, cold worked austenitic manganese steels, one containing nitrogen.

A high-nitrogen, high-chromium austenitic stainless steel (X5CrMnN1818K, $\leq 0.1\%$ C, 17 to 20% Mn, 17 to 20% Cr, $> 0.4\%$ N) is shown to be resistant to stress-corrosion cracking in environments where a high-carbon, low-chromium austenitic stainless steel (X55MnCr18K, 0.3 to 0.6% C, 17 to 20% Mn, 3 to 6% Cr) is highly susceptible to stress-corrosion cracking. In 90 °C water, the stress-corrosion resistance of the high-nitrogen austenitic steel is not affected by high nitrate contents, by high degrees of cold work, or by corresponding high yield strengths. The immunity to stress-corrosion cracking in 90 °C water is maintained even if the nitrogen content of the steel is as high as 1 wt%.

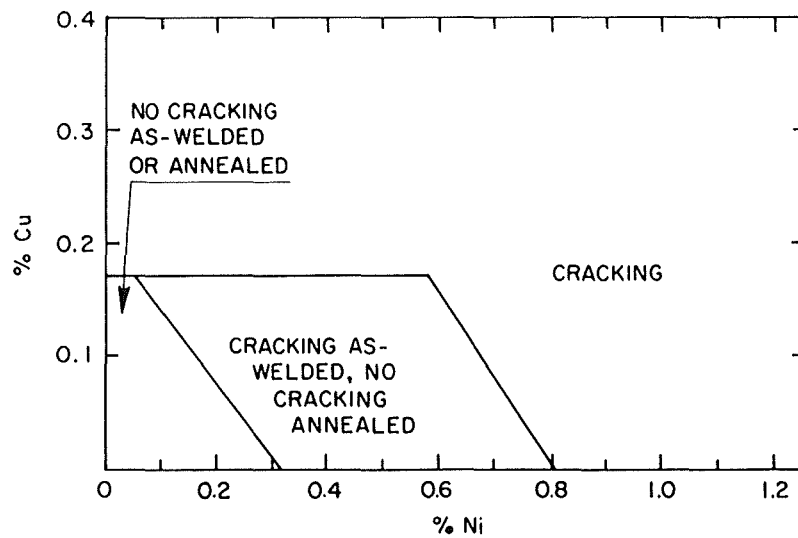
Stainless Steels: Effect of Partial Pressure and Temperature on Cracking

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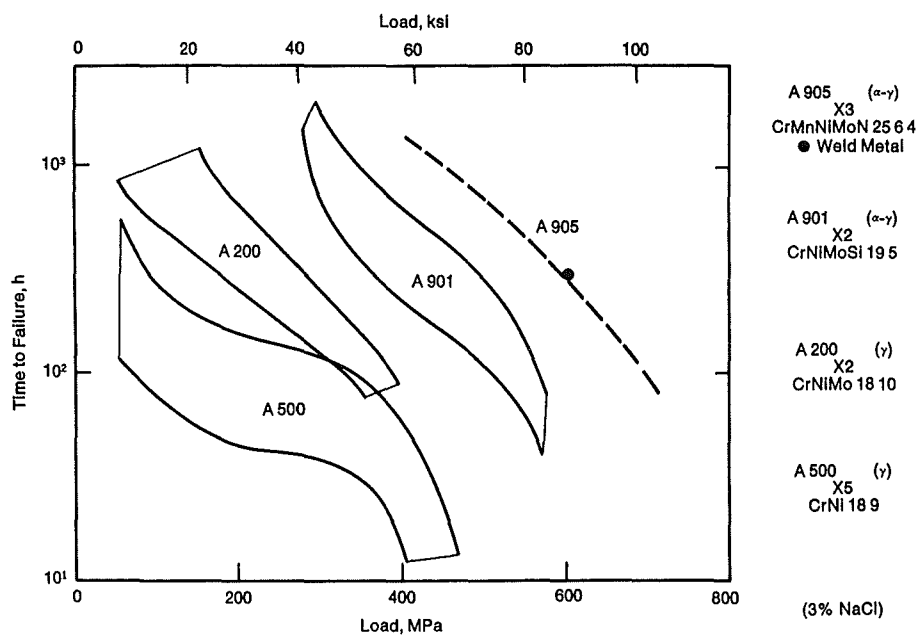


Effect of partial pressure and temperature on susceptibility to cracking of 9Cr-1Mo, 13Cr, and 25Cr steels. Open symbols, no failure; closed symbols, failure.

Stainless Steel: Effect of Copper and Nickel Contents on Cracking Resistance

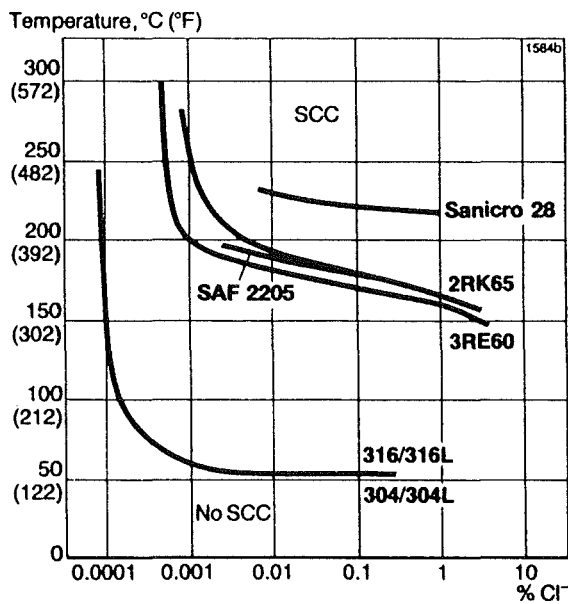


Effect of copper and nickel contents on the cracking resistance of U-bend specimens of ferritic Fe-18Cr-2Mo-0.35Ti-0.015C-0.015N stainless steels exposed to a boiling magnesium chloride solution at 140 °C.



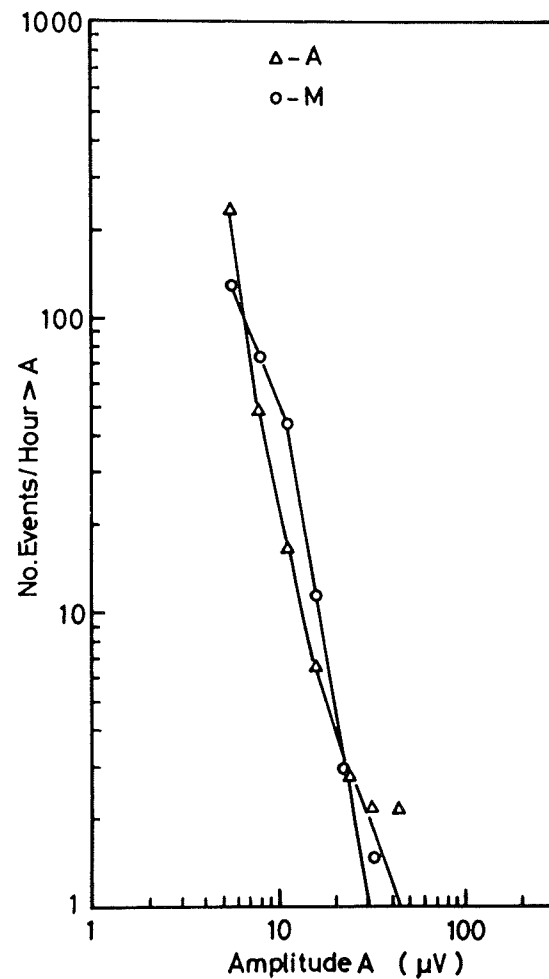
Comparison of stress-corrosion cracking resistance in 3% sodium chloride of austenitic and duplex stainless steels tested with crust formation.

Stainless Steel: Topside Applications and Stress-Corrosion Cracking



Practical experience with topside applications and laboratory results for stress-corrosion cracking in several stainless steels.

Source: B. Larsson, H. Gripenberg, and R. Mellstrom, Special Stainless Steels for Topside Equipment on Offshore Platforms, in Stainless Steels '84, The Institute of Metals, London, 1985, p 452-462.

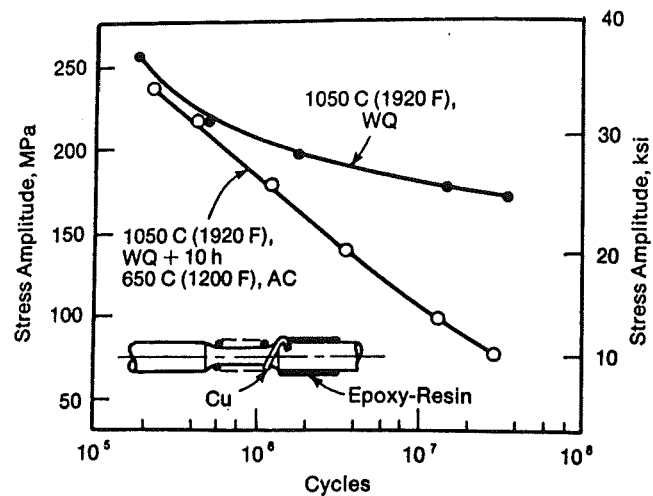


Cumulative event rate amplitude distributions for 140-kHz acoustic emission during stress-corrosion crack propagation. A = austenitic specimen; M = martensite-containing specimen.

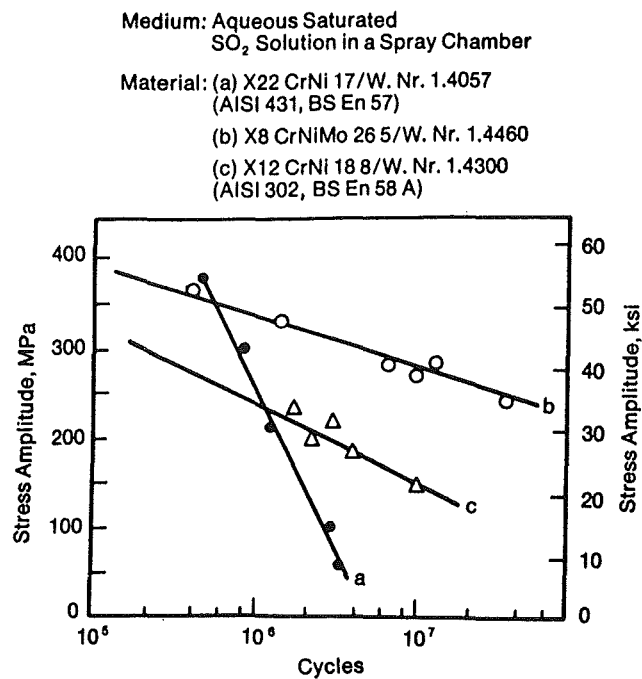
The specimens were tested in 42% magnesium chloride solution at 130 °C and 143 °C. Acoustic emission measurements indicated an active path corrosion mechanism in both austenitic and martensite-containing materials.

Stainless Steel: Effect of Sensitization on Corrosion Fatigue Behavior

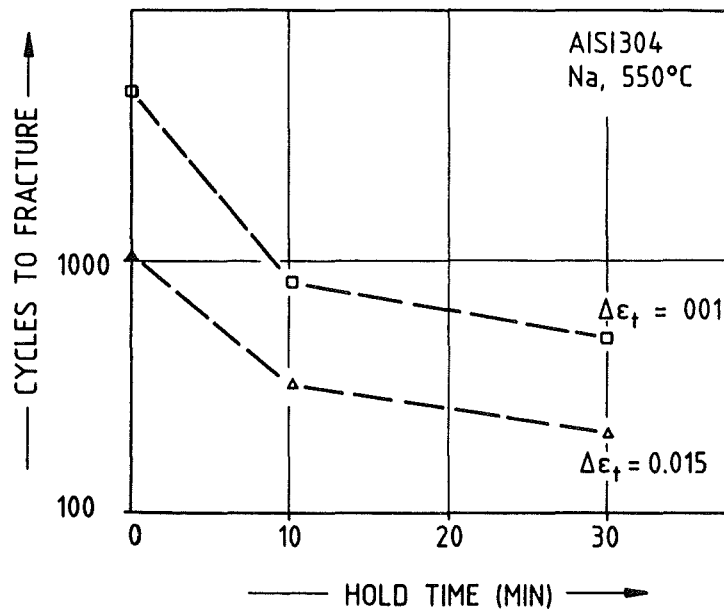
<u>Test Conditions:</u> <u>Material:</u> X12 CrNi 18 8/W. Nr. 1.4300 (AISI 302, BS En 58 A) Fatigue Limit (Rotating Bending) in Air: ± 285 MPa (41 ksi)	<u>Surface Treatment:</u> Grinding and Electrolytic Polishing <u>Frequency:</u> 2900 cpm <u>Solution:</u> Strauss Solution 40 C (105 F)
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Effect of sensitization on the corrosion fatigue strength of a stainless steel in Strauss solution at 40 °C.



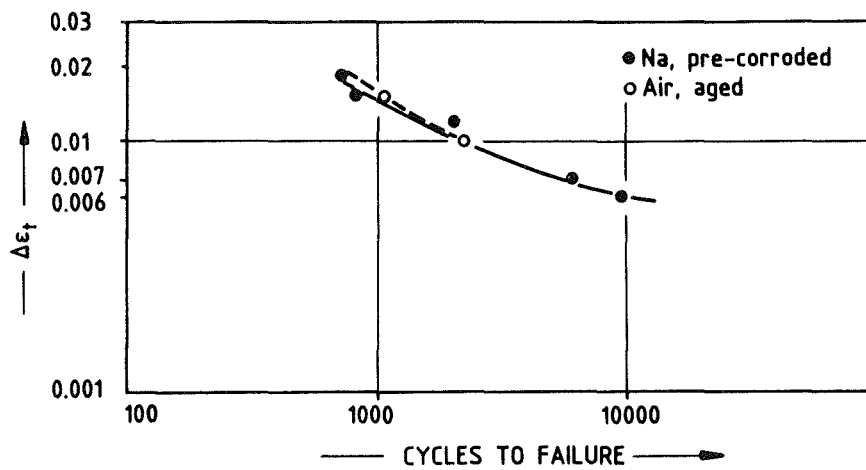
Corrosion fatigue strength of different types of stainless steels in saturated aqueous SO₂ solution at ambient temperature.



Effect of hold time at maximum tensile strain on the fatigue life of type 304 stainless steel in liquid sodium at 550 °C.

Type 304 Stainless Steel: Low-Cycle Fatigue Life in Liquid Sodium and Air

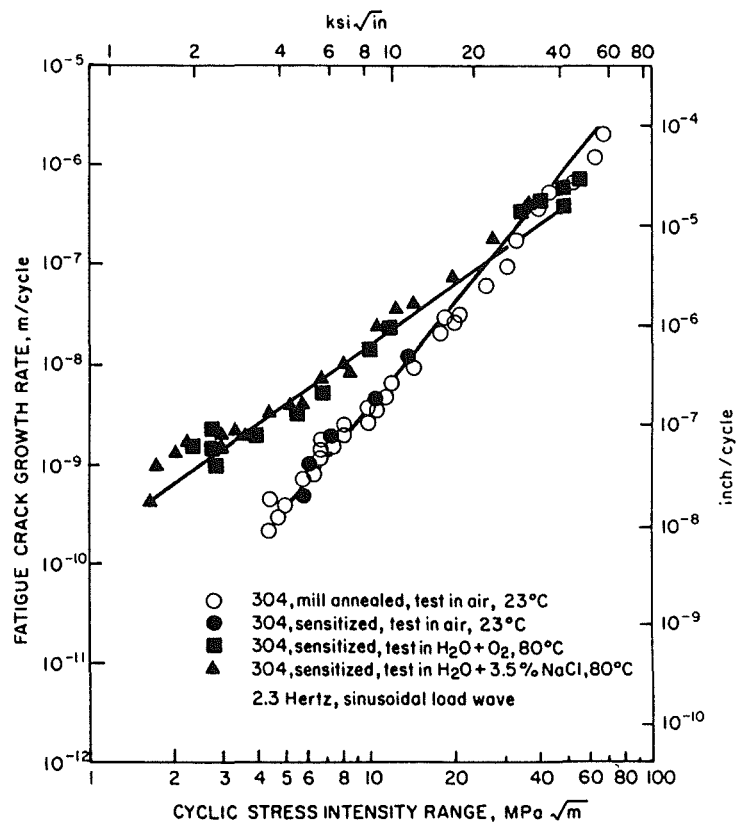
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Low-cycle fatigue life of type 304 stainless steel in liquid sodium and in air at 550 °C.

Specimens pre-corroded in decarburizing sodium suffered a slight reduction in fatigue life in sodium compared to specimens that received the same thermal treatment in air.

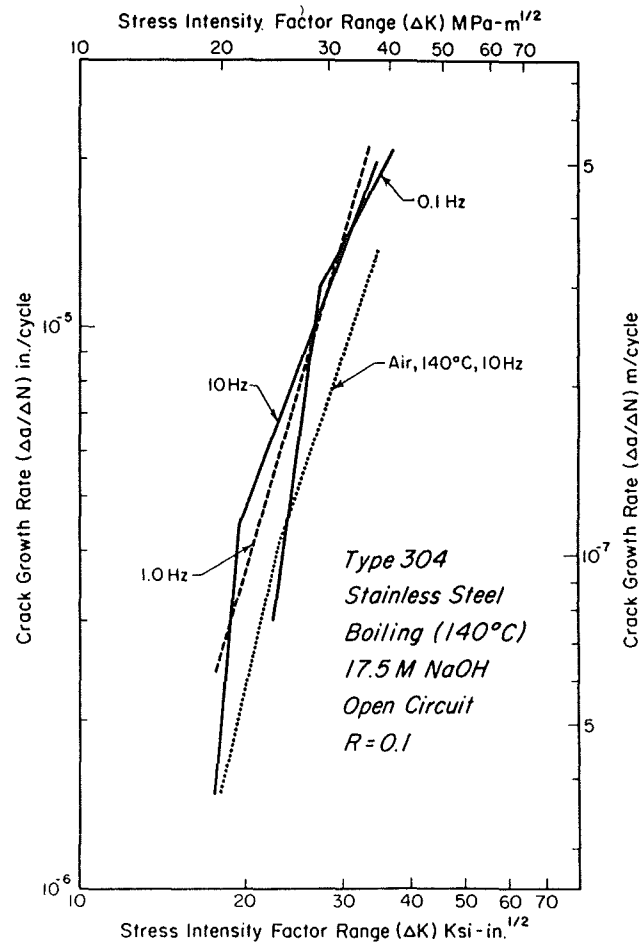
Type 304 Stainless Steel: Factors Affecting Growth Rate of Fatigue Cracks



Effect of environment and cyclic stress-intensity factor range on the growth rate of fatigue cracks in type 304 stainless steel.

Type 304 Stainless Steel: Fatigue Crack Growth Rate Curves in Sodium Hydroxide Solution

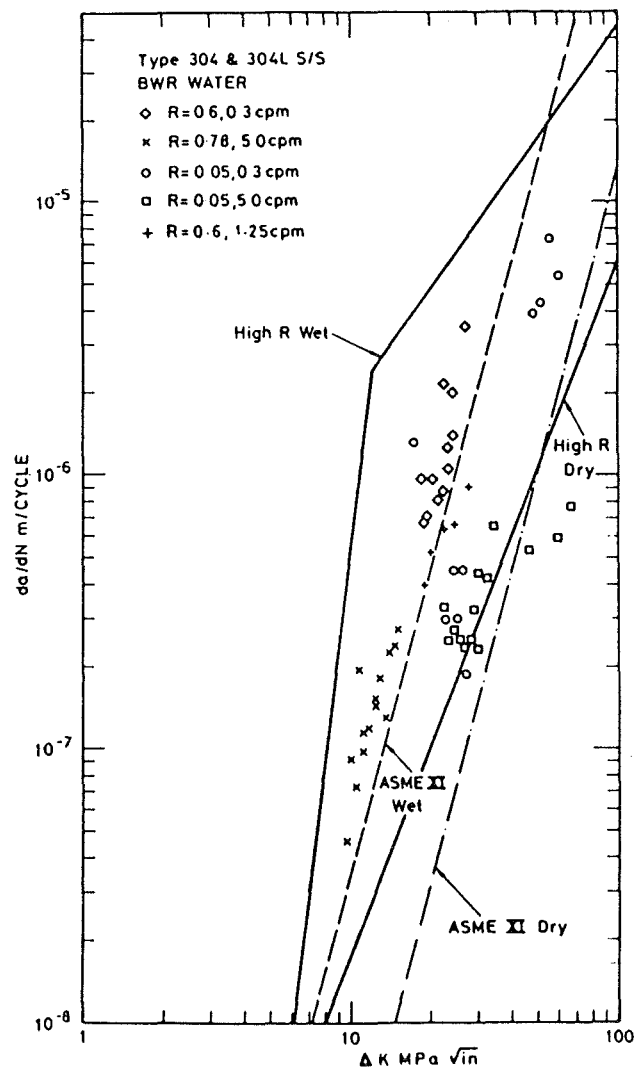
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Fatigue crack growth rate curves for type 304 stainless steel in boiling 17.5M sodium hydroxide solution showing effect of cyclic frequency.

Source: A. Boateng, J.A. Begley, and R.W. Staehle, Corrosion Fatigue of Type 304 Stainless Steel and Inconel 600 in a Boiling (140 °C) Concentrated Caustic Solution, Met. Trans. A, 14A, January 1983, p 67-74.

Austenitic Stainless Steels: Corrosion Fatigue Crack Growth Data

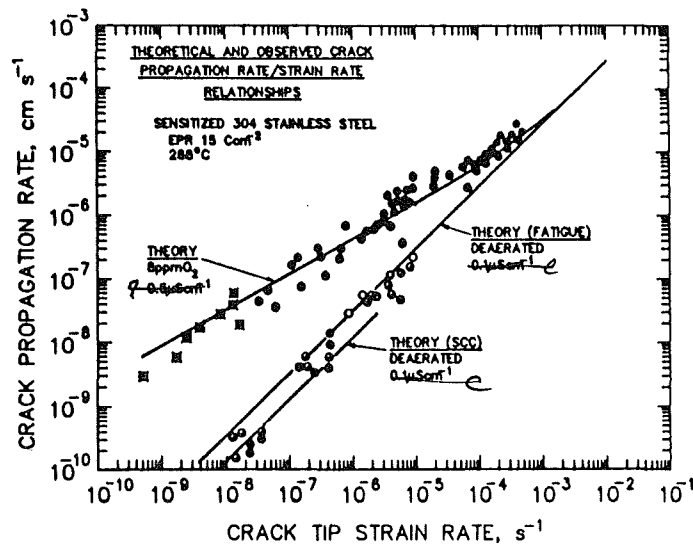


Corrosion fatigue crack growth data for two austenitic stainless steels in normal BWR water.

The solid and dashed lines are from Section 11 of the ASME Boiler and Pressure Vessel Code and are intended to represent the upper bounds for fatigue crack growth data for the conditions indicated. These upper bounds are modified periodically as more information becomes available.

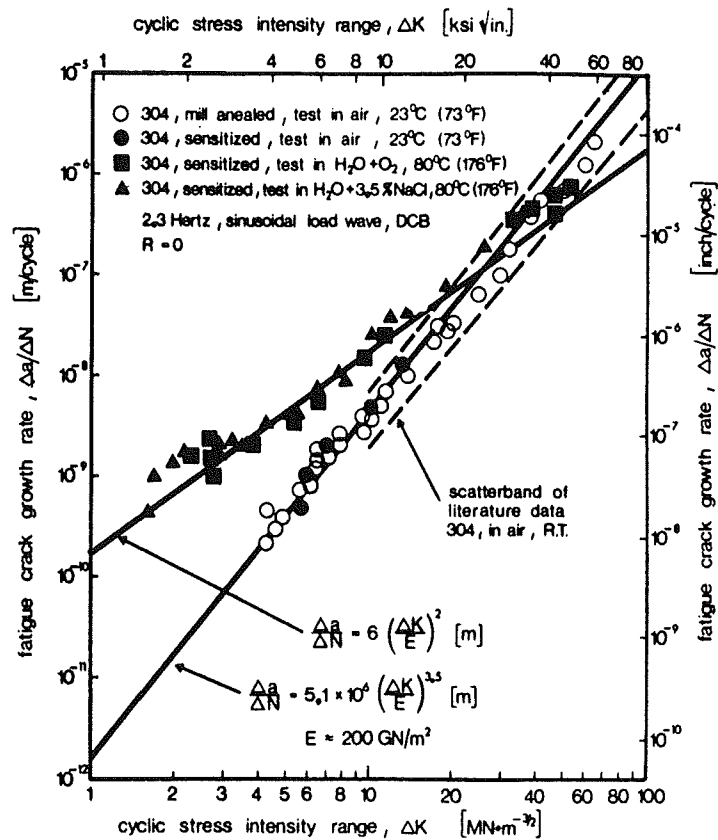
Type 304 Stainless Steel: Comparison between Observed and Theoretical Crack Propagation Rate and Crack Tip Strain Relationships

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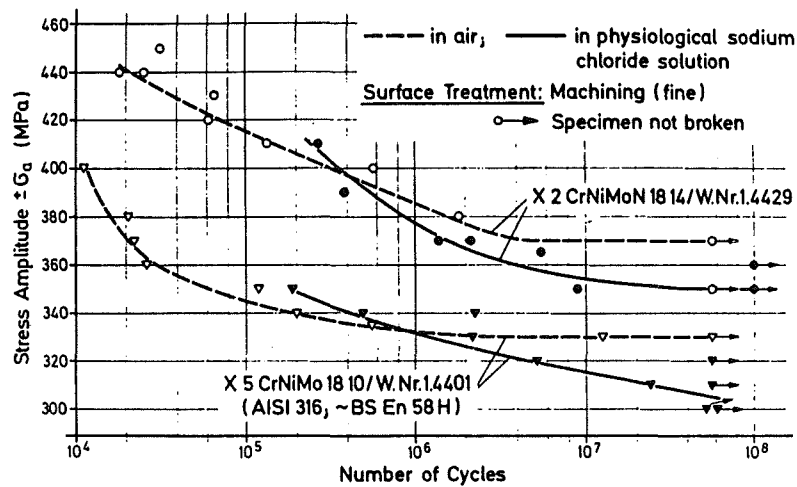


Comparison between observed and theoretical crack propagation rate and crack tip strain relationships for sensitized Type 304 stainless steel in 288 °C water, which is either aerated (8 ppm O₂) or deaerated.

Type 304 Stainless Steel: Fatigue Crack Growth Rates in Various Environments

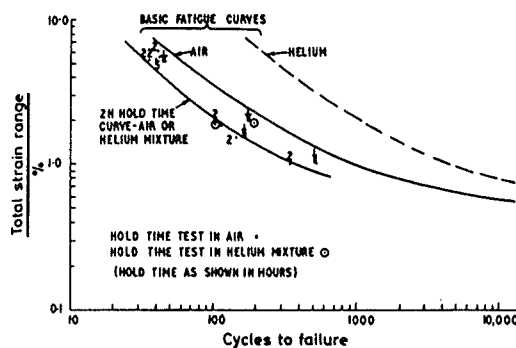


Rate of fatigue crack growth in various environments for type 304 stainless steel as a function of the range of the stress-intensity factor.



S-N curves for two austenitic stainless steels tested in air and in physiological sodium chloride solution.

Type 316 Stainless Steel: Effect of Environment on Creep/Fatigue Behavior

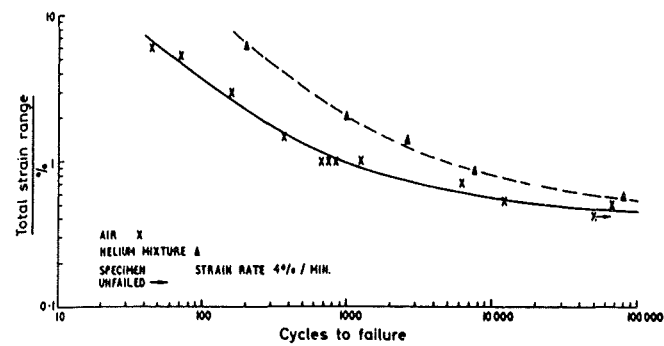


Effect of environment on creep/fatigue behavior of type 316 stainless steel at 625 °C.

The helium contained 200×10^{-6} atm H_2 and 10^{-6} atm H_2O . Well-defined striations were present on the fracture surface of specimens tested in air, but were not present on the fracture surfaces of specimens tested in helium or in air with a 2-h hold time.

Type 316 Stainless Steel: Effect of Environment on High-Strain Fatigue Behavior

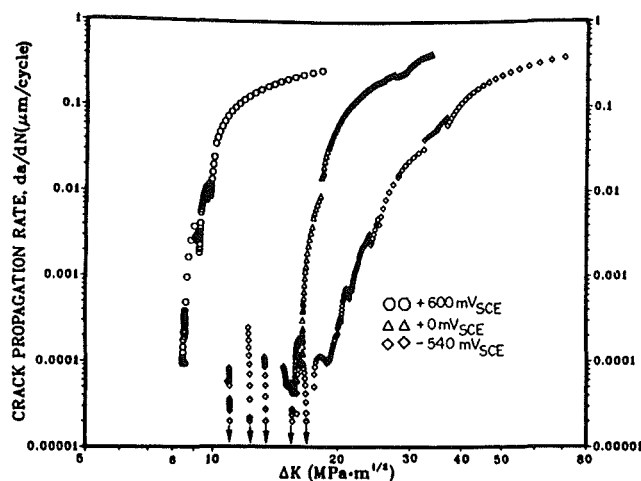
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Effect of environment on high-strain fatigue behavior of type 316 stainless steel at 625 °C.

Source: D.S. Wood, et al., Preliminary Results of Effect of Environment on the Low Cycle Fatigue Behaviour of Type 316 Stainless Steel and 9% Cr Ferritic Steel, in The Influence of Environment on Fatigue, Institution of Mechanical Engineers, London, 1977, p 11-20.

Type 316L Stainless Steel: Fatigue Crack Propagation Behavior

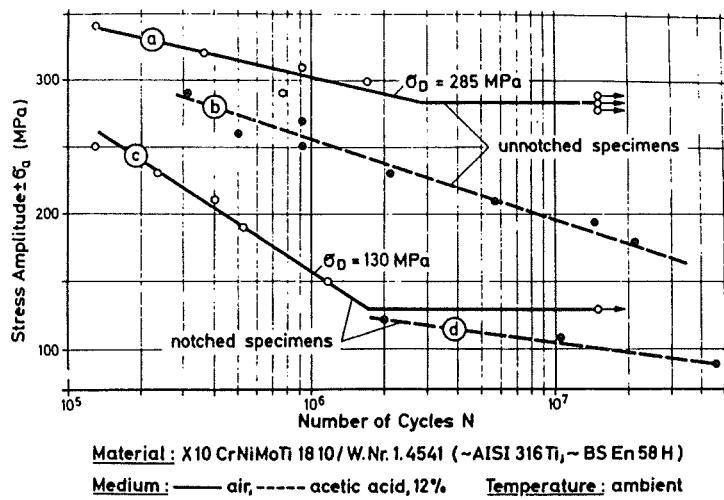


Potentiostatically controlled fatigue crack propagation behavior of type 316L stainless steel in 1M sodium chloride. Arrows indicate crack arrest.

The fatigue crack growth tests were carried out at a frequency of 123 Hz, a frequency high enough to maintain the composition of the crack solution close to that of the bulk solution. Roughness-induced crack closure was minimized at high anodic potentials by electrochemical erosion. Thermodynamic considerations showed that hydrogen played no role in fatigue crack propagation. Anodic dissolution processes were also considered to have a minor effect on the rate of fatigue crack propagation. A model was proposed for stage I fatigue crack growth based on the effect of oxide nucleation rate on restricted slip reversal.

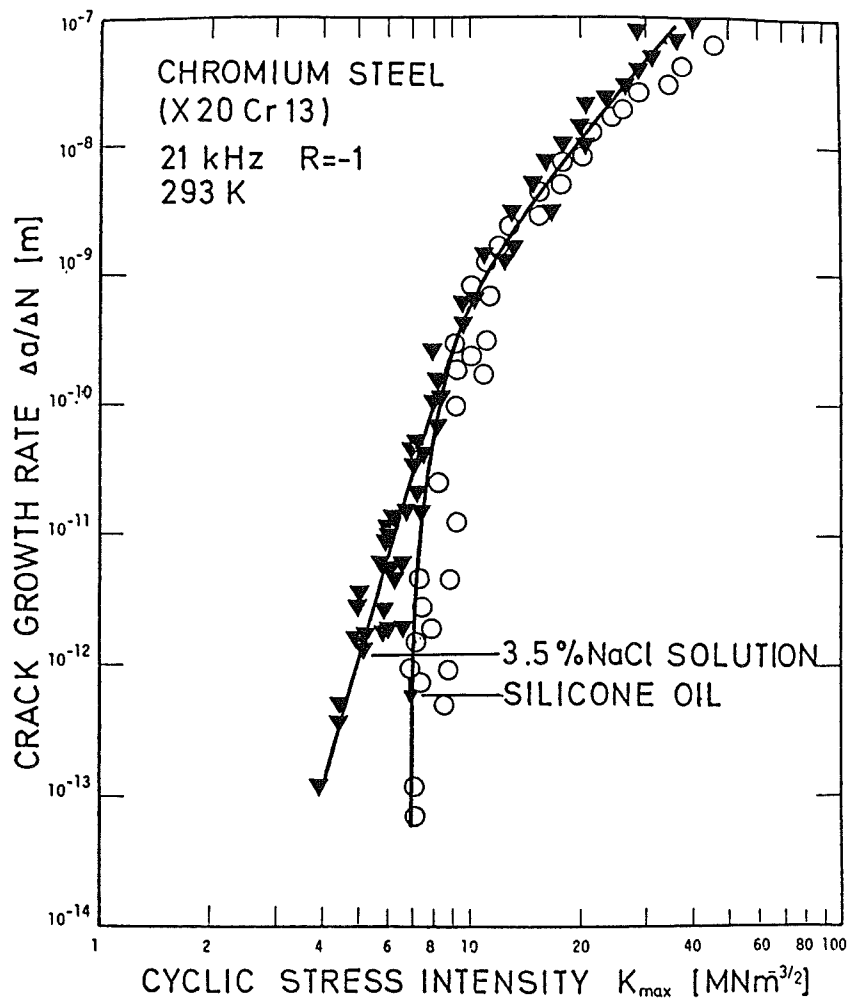
Type 316 Stainless Steel: S-N Curves for Notched and Unnotched Specimens

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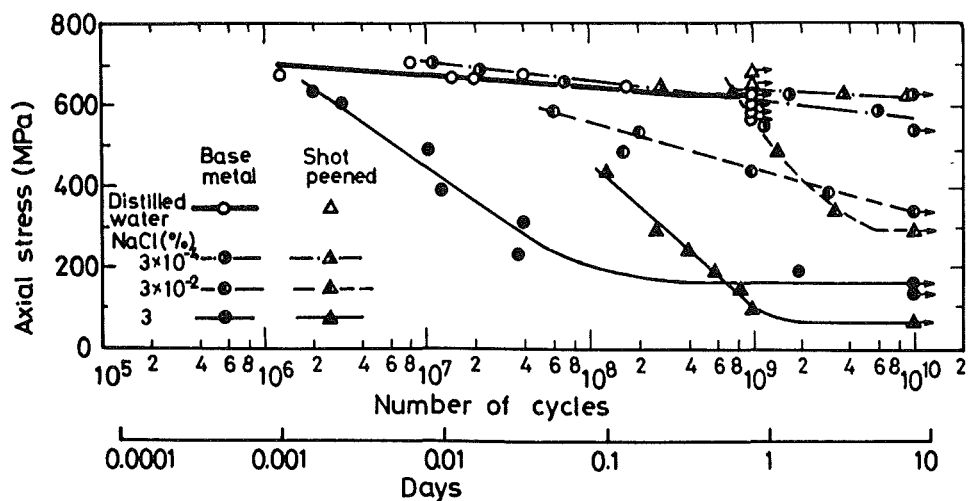
S-N curves for notched and unnotched specimens of type 316 stainless steel tested in air and in 12% acetic acid.

Chromium Steel: Effect of Environment on Crack Growth Rates



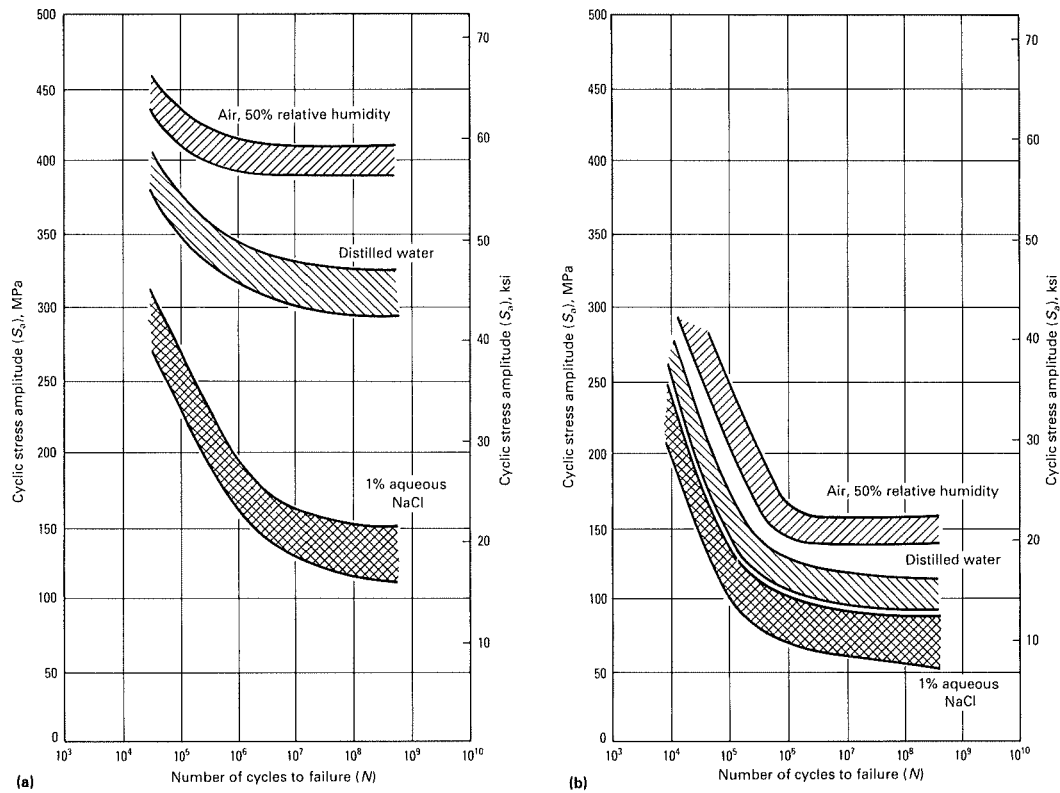
Effect of the environment on the crack growth rates in a chromium steel as a function of the cyclic stress-intensity.

In the 3.5% sodium chloride solution, no well-defined threshold level was found despite the extremely low rates of fatigue crack propagation studied. This finding is consistent with the lack of a well-defined endurance limit for steels in corrosive environments.

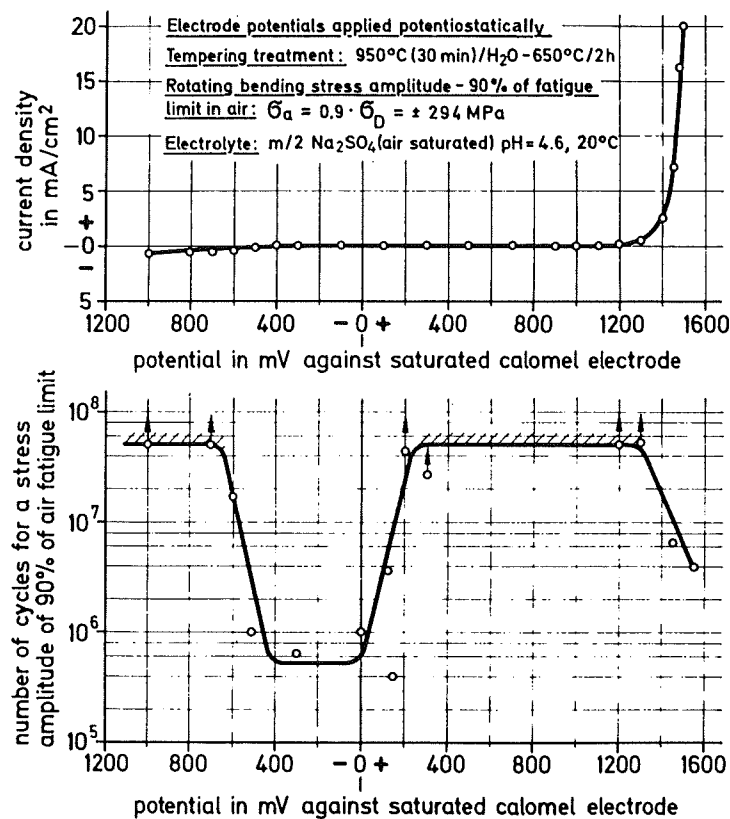


S-N curves for a 13Cr stainless steel tested at 16 kHz in distilled water and in sodium chloride solutions of differing concentrations.

13Cr Steel: Corrosion Fatigue Data



Corrosion fatigue data for specimens of 13Cr steel. Rotating bending tested (mean load zero) at a frequency of 50 Hz and temperature of 23 °C (73 °F). (a) Smooth specimens. (b) Notched specimens. Note that because the notch drastically lowers the fatigue curve for air the range of corrosion fatigue results for notched specimens is thereby compressed.



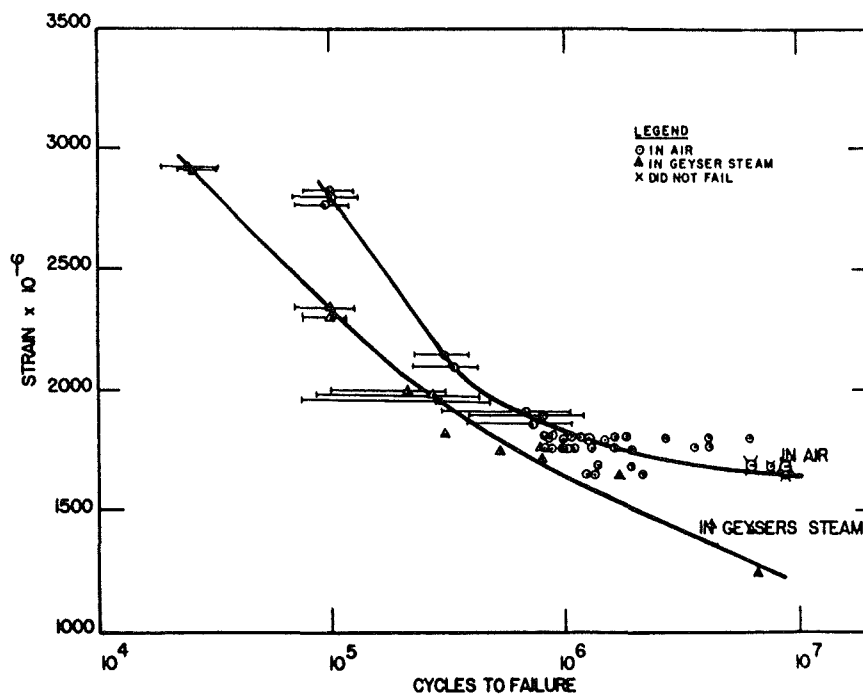
Corrosion fatigue behavior of a tempered 13% Cr steel.

Under certain conditions, steel can be effectively protected against corrosion fatigue by cathodic polarization. In acid solutions, cathodic protection cannot be applied if the protective cathodic current density becomes too high and hydrogen is evolved. In weakly acid, neutral, and alkaline electrolytes, cathodic protection against corrosion is excellent. At a pH of 4.6, full cathodic protection is achieved even at amplitudes as high as 90% of the air fatigue limit, provided the potential is lower than -850 mV_{SCE}.

Geothermal Turbines: Corrosion Fatigue of Turbine Buckets and Stress-Corrosion Cracking of Monoblock Rotors in Geothermal Steam

MECHANICAL PROPERTIES OF TEST MATERIALS

Alloy	Heat Treatment	Austenitizing Temp., °C	Tempering Temp., °C	Tempering Time, Mins.	R Hardness	Fatigue Strength, MPa, 10^7 Cycles		
						Geysers Air	Clean Steam	Geysers Steam
403	2	990	650	60	23	317	241	Not Tested
403	3	990	650	80	24	317	262	Not Tested
422	7	1040	760	120	22	317	227	282
403	8	1040	680	60	21	310	269	

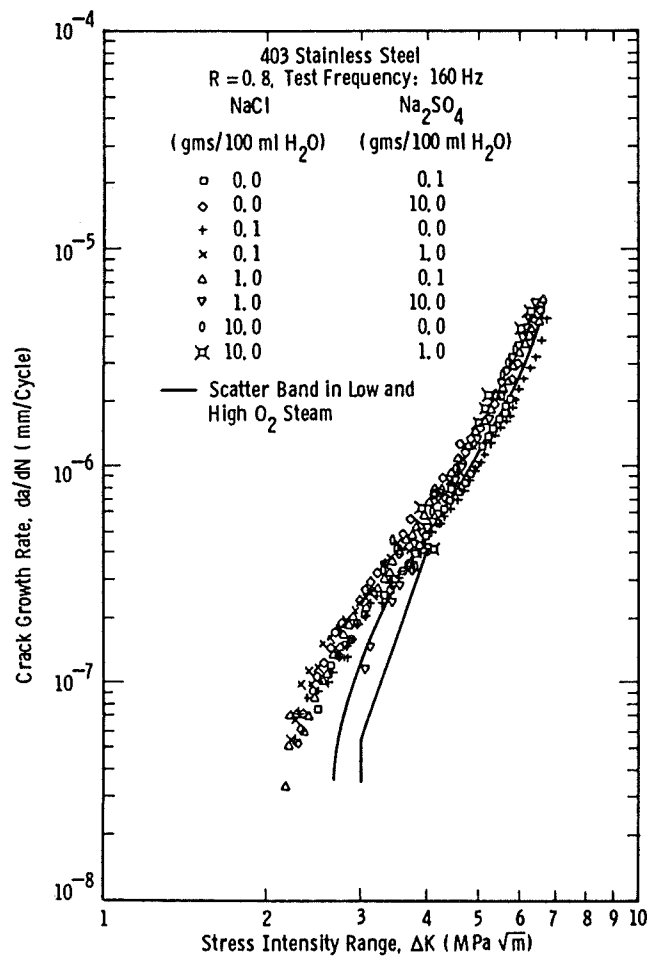


Strain versus cycles-to-failure for type 403 stainless steel heat treated to 23 HRC.

Rectangular specimens with a central hole were tested in a reverse bending machine. Steam temperature and pressure in the test chambers were adjusted to simulate that found in turbine stages. The tests in air were performed at room temperature. The accompanying table provides information on the effect of heat treatment on fatigue strength.

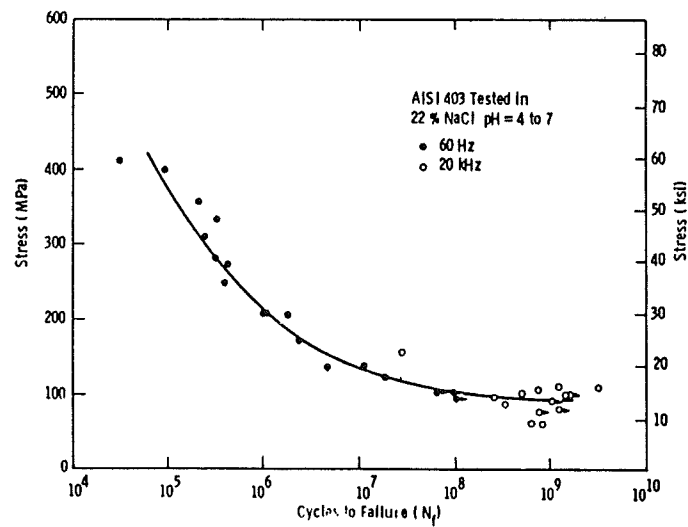
Type 403 Stainless Steel: Near-Threshold Corrosion Fatigue Crack Propagation Data

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Near-threshold corrosion fatigue crack propagation data for type 403 stainless steel obtained in NaCl/Na₂SO₄ solution.

Type 403 Stainless Steel: Conventional and Ultrasonic Fatigue Strength

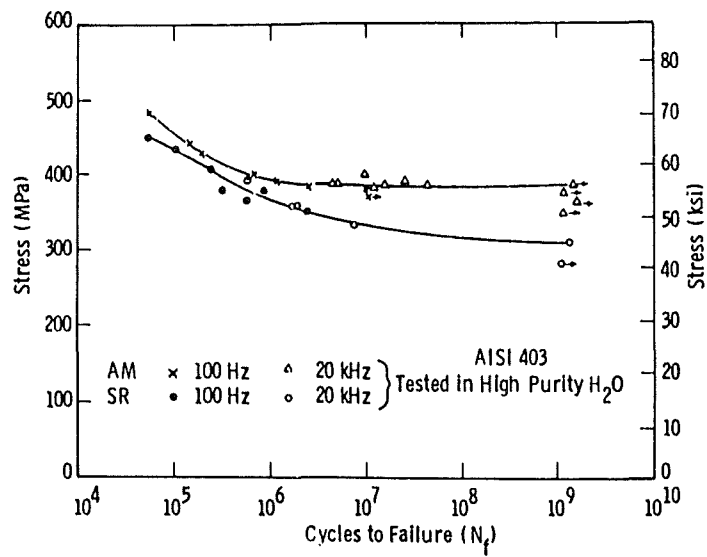


Comparison of conventional and ultrasonic fatigue strength of AISI 403 stainless steel in 22% sodium chloride salt solution.

Source: L.E. Willert, T.M. Rust, and V.P. Swaminathan, High and Low Frequency Corrosion Fatigue of Some Steam Turbine Blade Alloys, in Ultrasonic Fatigue, J.M. Wells, et al., Ed., American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, PA, 1982, p 333-348.

Type 403 Stainless Steel: Fatigue Results in Pure Water

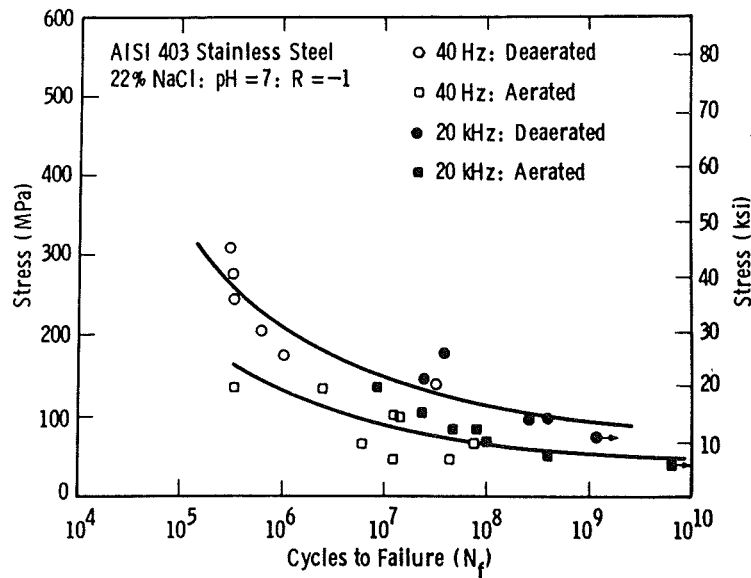
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Comparison of fatigue results for AISI 403 stainless steel tested in pure water by conventional and ultrasonic means (AM, as machined; SR, stress relieved).

Source: L.E. Willert, T.M. Rust, and V.P. Swaminathan, High and Low Frequency Corrosion Fatigue of Some Steam Turbine Blade Alloys, in Ultrasonic Fatigue, J.M. Wells, et al., Ed., American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, PA, 1982, p 333-348.

Type 403 Stainless Steel: Comparison of Fatigue Data in Sodium Chloride

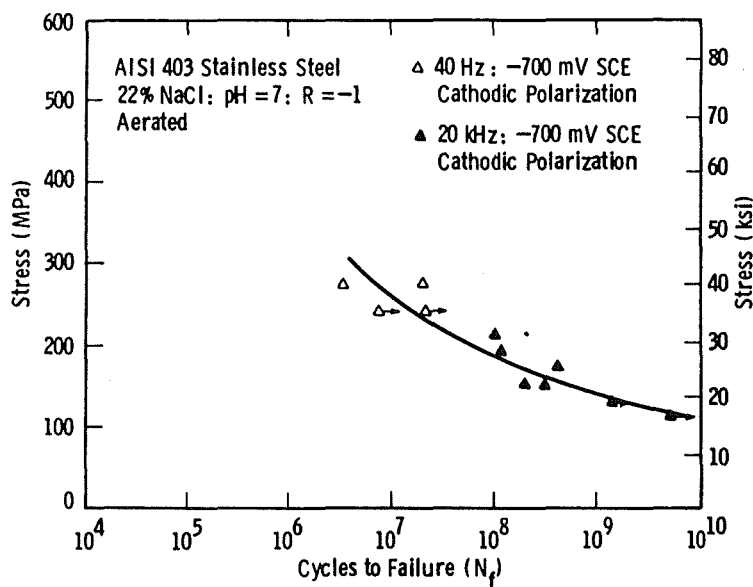


Comparison of fatigue data for AISI 403 stainless steel for tests at 40 Hz and 20 kHz in aerated and deaerated 22% sodium chloride solutions.

For a given set of test conditions, the S-N data and the crack initiation morphologies were similar at the two frequencies. Cracks initiated at the base of pits formed at both frequencies, even though some of the tests at 20 kHz lasted only a few hundred seconds. It was proposed that the similarity was due to a rate-limiting step involving the transport of metal ions out of the pits. This transport occurs at a high rate in ultrasonic fatigue.

Type 403 Stainless Steel: Comparison of Fatigue Data as a Function of Frequency

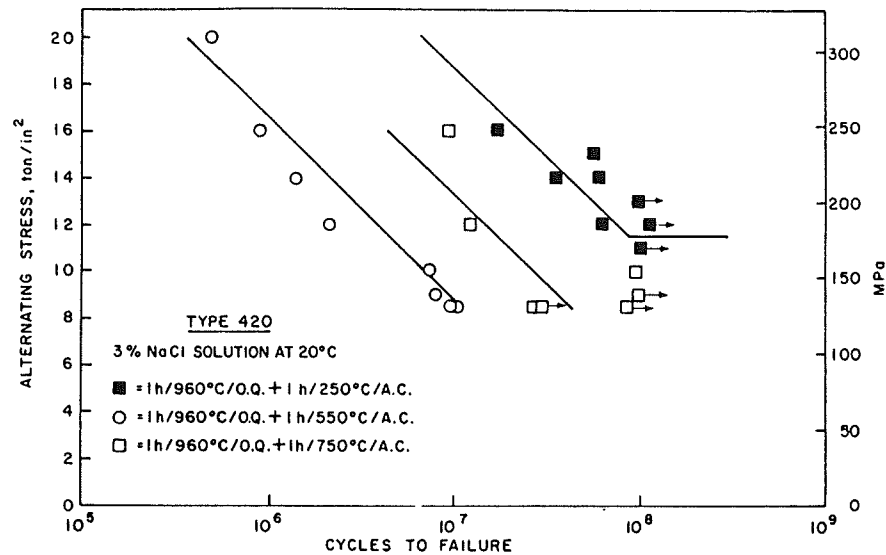
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Comparison of fatigue data for tests at 40 Hz and 20 kHz with cathodic polarization in aerated 22% sodium chloride solution.

Note that at 20-kHz fatigue lives of up to 10^{10} cycles were determined and that cathodic polarization has resulted in an improvement in resistance to corrosion fatigue.

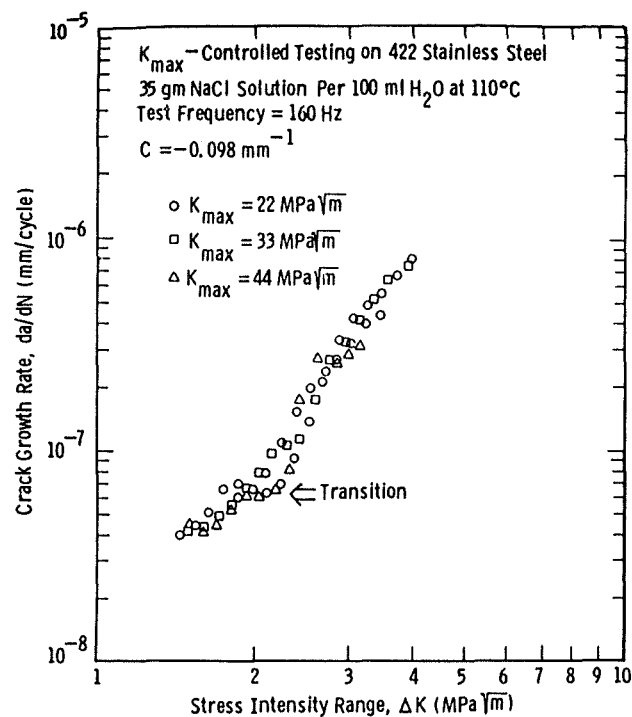
Type 420 Stainless Steel: Effect of Tempering Temperature on Corrosion Fatigue, in NaCl



Fatigue behavior of notched and unnotched rotating bend specimens of a titanium-stabilized type 316 stainless steel in air and in acetic acid.

Type 422 Stainless Steel: Near-Threshold Fatigue Crack Growth Rates

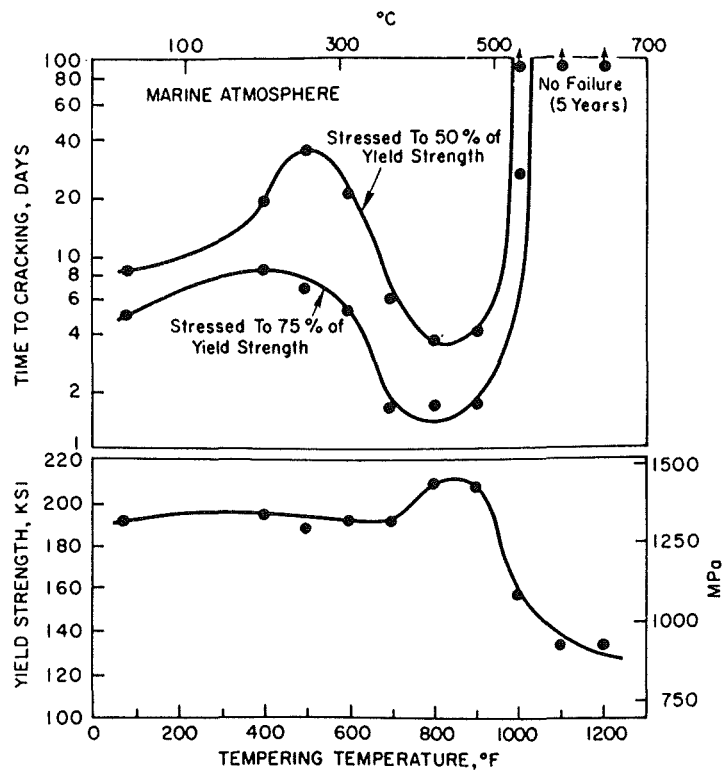
251



Effect of K_{max} values on near-threshold fatigue crack growth ratios of type 422 stainless steel in a boiling sodium chloride solution.

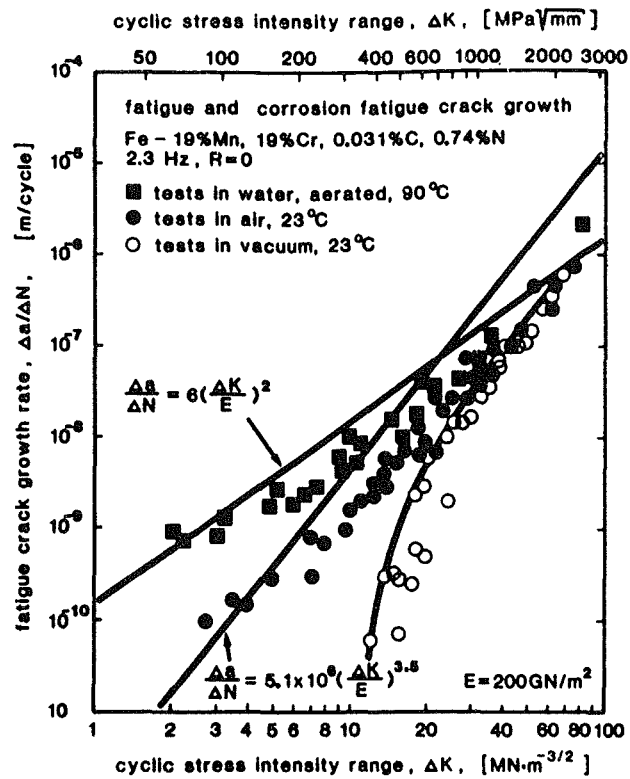
The transition illustrated above involved a transition from corrosion fatigue to stress-corrosion cracking. It occurred at very high load ratios ($R > 0.9$), and at very low ΔK levels ($< 2.1 \text{ MPa}\sqrt{m}$).

USS 12MoV Stainless Steel: Effect of Tempering Temperature on Cracking Resistance and Yield Strength



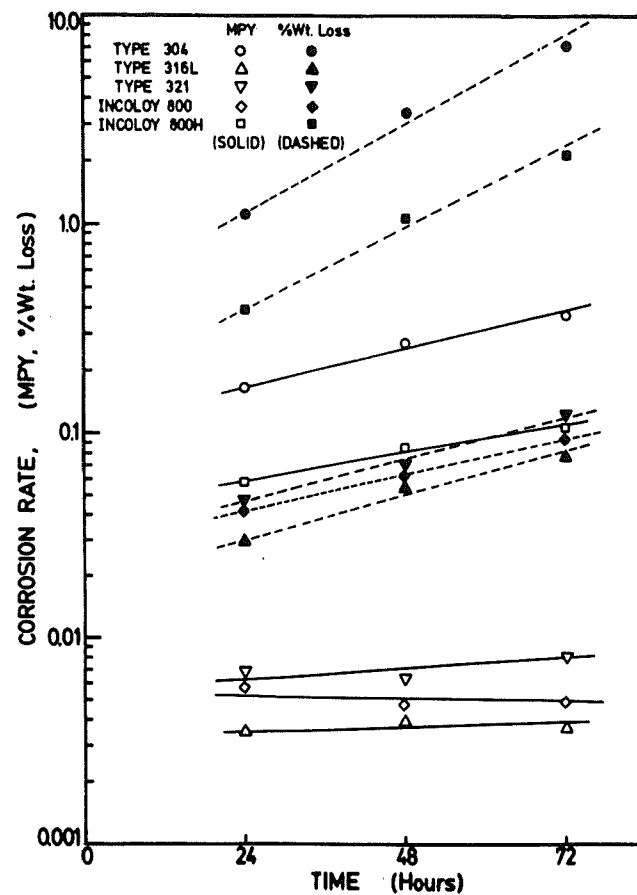
Effect of tempering temperature on the cracking resistance and yield strength of USS 12MoV stainless steel.

The minimum resistance to stress-corrosion cracking is exhibited in the tempering range wherein alloy carbides are precipitated. Above this temperature range, the resistance increases rapidly as the strength decreases.



Effect of cyclic stress-intensity range and various environments on the growth rate of fatigue cracks in retaining ring steel 18/8. Note that water accelerates crack growth rates markedly, especially at low cyclic stress-intensity ranges.

Austenitic Stainless Steels: Comparison of Corrosion Behavior



Comparison of corrosion behavior of austenitic stainless steels subjected to low-cycle fatigue at 650 °C, as measured by percent weight loss and mils per year loss (MPY) as a function of time of exposure to Strauss test conditions.

Low-cycle fatigue at 650 °C promoted the precipitation of chrome carbides, particularly in type 304 stainless steel and Incoloy 800H. The rate of weight loss after low-cycle fatigue was twice as high as for material that had been sensitized at 650 °C for a time equal to the average time of exposure to 650 °C during the low-cycle fatigue tests.

ALUMINUM ALLOYS

ALUMINUM ALLOYS

Only aluminum alloys that contain appreciable amounts of soluble alloying elements—primarily copper, magnesium, silicon, and zinc—are susceptible to stress-corrosion cracking (SCC). For most commercial alloys, tempers have been developed to improve resistance to SCC. For this class of alloys, electrochemical considerations have provided a basis for developing aluminum alloys and tempers resistant to SCC.

SCC in aluminum alloys is characteristically intergranular, and according to the electrochemical theory this requires a conditions along grain boundaries that makes anodic to the rest of the microstructure. The most anodic regions may be either the precipitates formed in grain boundaries, or the regions adjoining the boundaries that have been depleted of solute. In addition, there is recent evidence that hydrogen generated at crack tips in aqueous solutions or moist air also may be playing in the fracture mechanism.

Fatigue strengths of aluminum alloys are lower in such corrosive environments as seawater and other salt solutions than in air—especially when evaluated by low-stress, long duration tests. Such corrosive environments produce smaller reductions in fatigue strength in the more corrosion resistant alloys such as the 5xxx and 6xxx series, than in the less resistant alloys such as the 2xxx and 7xxx series.

As with stress-corrosion cracking of aluminum alloys, corrosion fatigue is promoted by the presence of water. In contrast to SCC however, corrosion fatigue is not as appreciably affected by test direction because the fracture resulting from this type of attack is predominantly transgranular.

Basic Temper Designations

- T3** solution heat-treated, cold worked, and naturally aged to a substantially stable condition.
Applies to products which are cold worked to improve strength after solution heat-treatment, or in which the effect of cold work in flattening or straightening is recognized in mechanical property limits.
- T4** solution heat-treated and naturally aged to a substantially stable condition.
Applies to products which are not cold worked after solution heat-treatment, or in which the effect of cold work in flattening or straightening may not be recognized in mechanical property limits.
- T6** solution heat-treated and then artificially aged.
Applies to products which are not cold worked after solution heat-treatment, or in which the effect of straightening may not be recognized in mechanical property limits.
- T651** applies to plate stretched to give a 1-1/2 to 3% permanent set after solution heat-treatment.
- T7** solution treated and then overaged/stabilized.
Applies to wrought products that are artificially aged after solution heat-treatment, to carry them beyond a point of maximum strength to provide control of some significant characteristic. Also applies to cast products that are artificially aged after heat-treatment to provide dimensional and strength stability.
- T7351** Applies to plate stretched to give 1-1/2 to 3% permanent set after solution heat-treatment and then overaged for resistance to SCC.

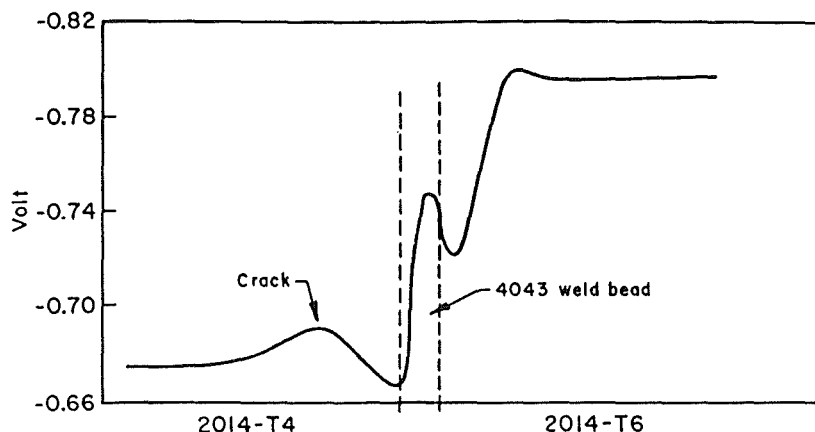
O annealed.

H strain-hardened (wrought products only).

Applies to products which have their strength increased by strain hardening, with or without supplementary thermal treatments, to produce some reduction in strength. The H is always followed by two or more digits.

W solution heat-treated.

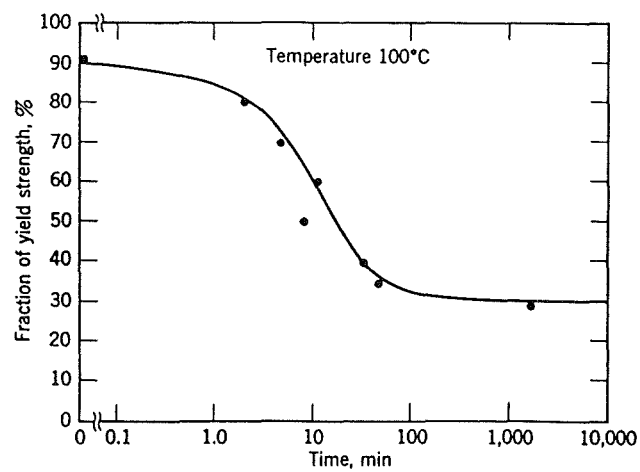
An unstable temper applicable to products which spontaneously age at room temperature after solution heat-treatment.



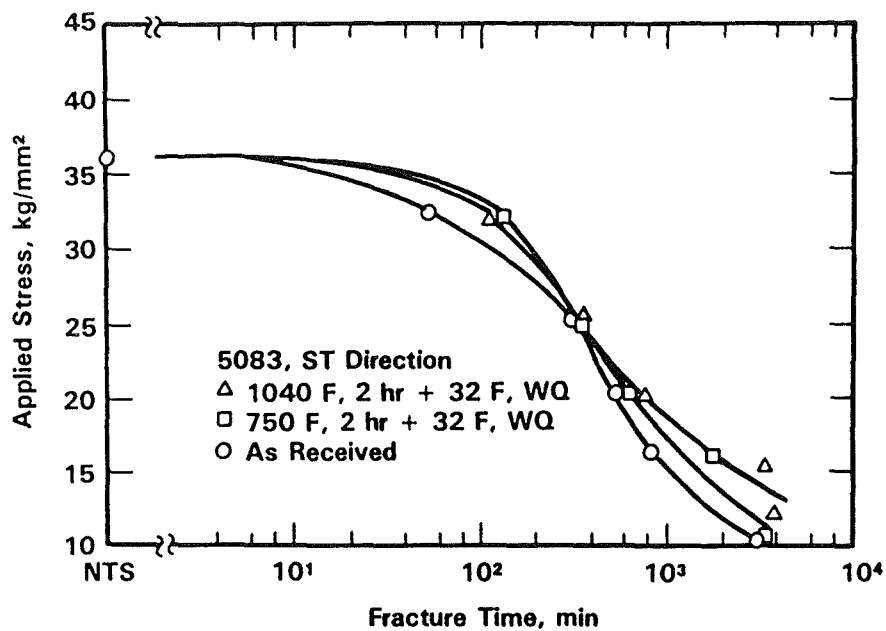
Changes in solution potential across a weld zone between 2014-T4 and 2014-T6.

The changes were due to the heat of welding. The potentials were measured against a 0.1N calomel reference electrode in an electrolyte containing 53 g/L NaCl and 3 g/L H₂O₂. The welded specimen was tested in bending at 80% of the joint strength and exposed to 3.5% NaCl solution by alternate immersion. Failure occurred in the heat-affected zone where the heat of welding produced a metallurgical condition that was susceptible to stress-corrosion cracking.

2024-T4 Aluminum Alloy: Time-to-Failure as a Function of Statically Applied Stress

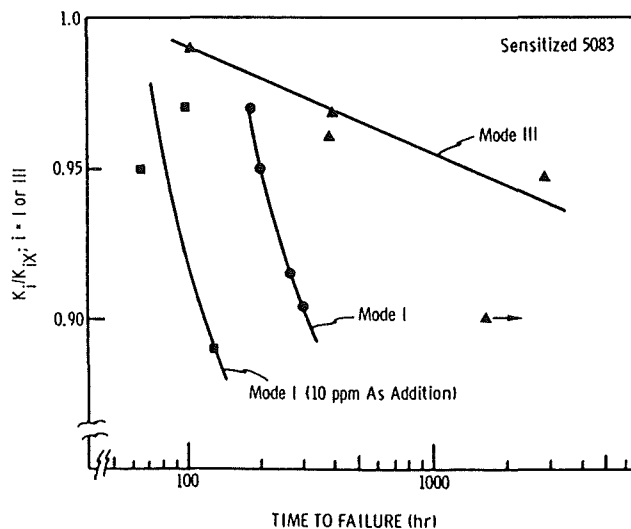


Time-to-failure of 2024-T4 aluminum alloy wetted with mercury amalgam at 100 °C as a function of the statically applied stress.



Stress-corrosion cracking curves in 3.5% NaCl + 0.5% H₂O₂ solution for the short-transverse (ST) direction specimens of 5083 aluminum heat treated at various solution temperatures.

5083 H131 Aluminum Alloy: Normalized Stress-Intensity Versus Time-to-Failure

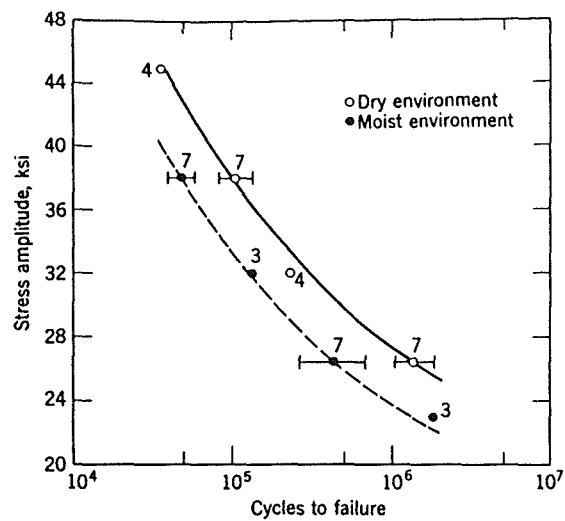


Normalized stress-intensity versus time-to-failure in Modes I and III for sensitized 5083 H131.

The susceptibility was evaluated in Mode I (tension) and Mode III (torsion), using precracked fracture toughness specimens. The ordinate is normalized by dividing the applied stress-intensity by the corresponding fracture toughness. The environment was 0.6M (3.5 wt%) NaCl + 0.02M Na₂Cr₂O₇ + 0.07M NaC₂H₃O₂ + HC₂H₃O₂ at a pH of 4. This solution inhibits general corrosion and simulates the conditions for stress-corrosion cracking in seawater. Arsenic, a hydrogen recombinant inhibitor, was added to the solution for additional tests in Mode I. The greater susceptibility found in Mode I was taken to indicate that hydrogen embrittlement was more dominant than dissolution for this loading mode.

6061-T6 Aluminum Alloy: S/N Curves for Unnotched Specimens in High and Low Humidity Environments

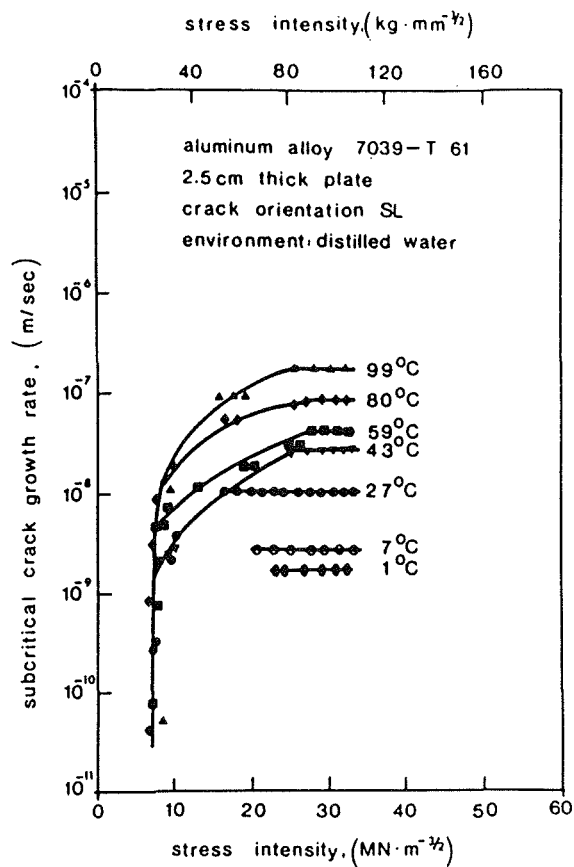
263



S/N curves for unnotched specimens of aluminum alloy 6061-T6 in high and low humidity environments ($R = -1/2$).

Numbers adjacent to symbols indicate the number of replicate tests. Horizontal bars indicate the range of ± 1 standard deviation.

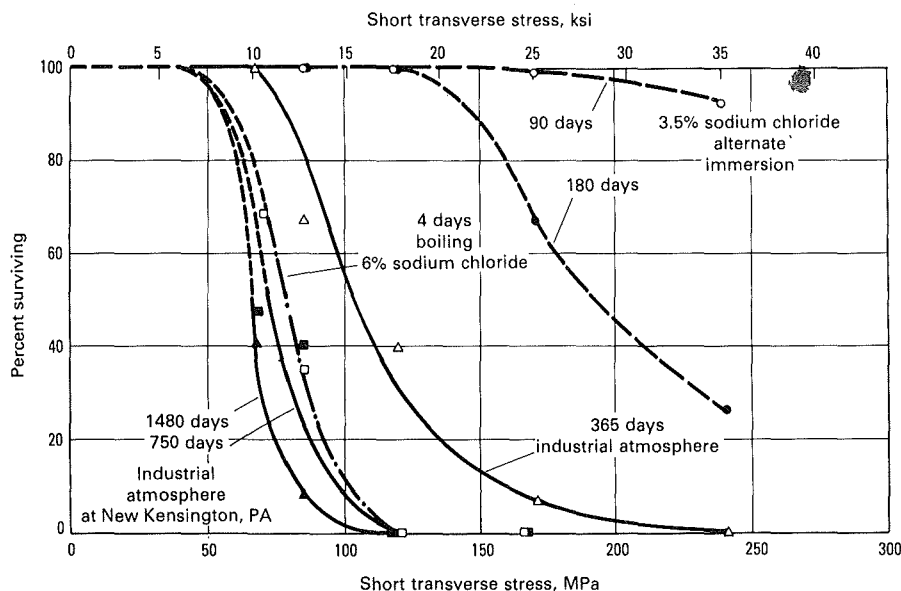
7039-T61 Aluminum Alloy: Effect of Temperature and Stress-Intensity on the Subcritical Crack Growth Rate, in Distilled Water



Effect of temperature and stress-intensity on the subcritical crack growth rate in an aluminum alloy exposed to distilled water.

7039-T64 Aluminum Alloy Rolled Plate: Results of Accelerated Corrosion Tests, in Sodium Chloride

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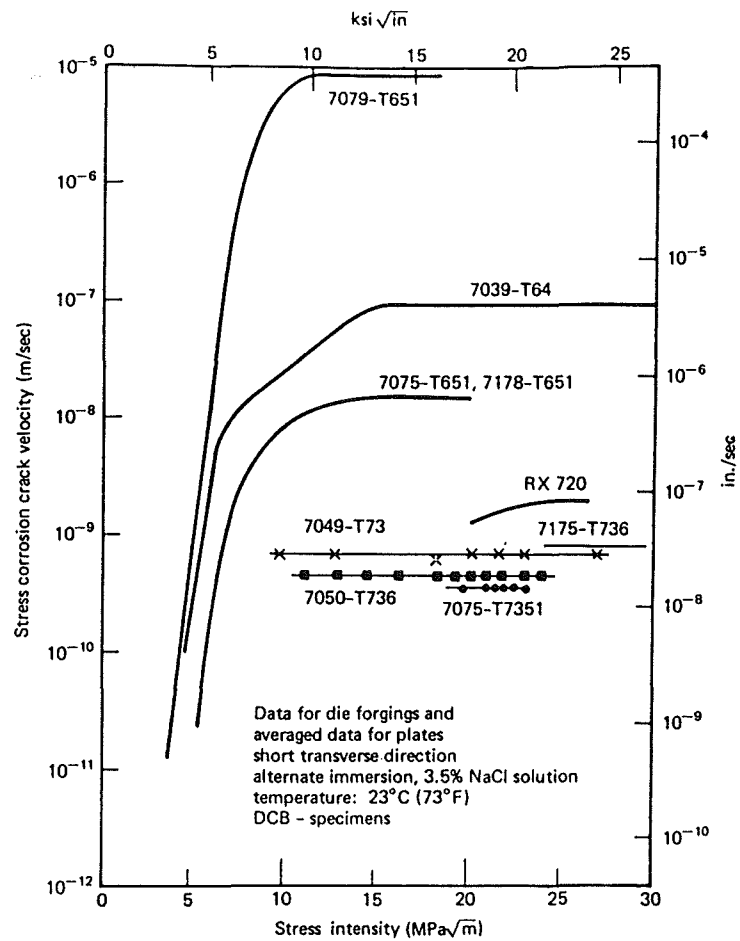


Correlation of results of accelerated tests with service environment (industrial atmosphere). The results are for five lots of rolled plate of the aluminum alloy 7039-T64 (4.0Zn-2.8Mg-0.3Mn-0.2Cr). The tests in 3.5% sodium chloride were similar to ASTM G 44, except that the salt solution was made with commercial grade sodium chloride and New Kensington tap water.

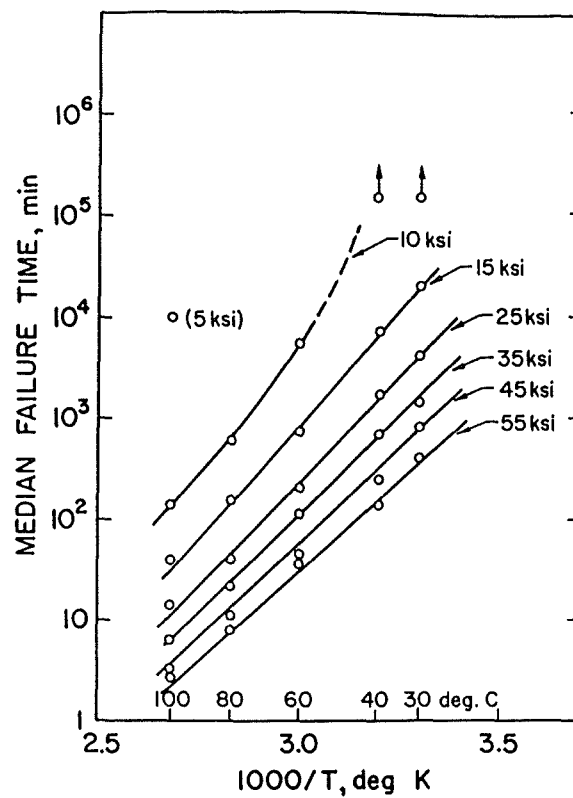
A 4-day exposure by continuous immersion in boiling 6% sodium chloride solution is widely used by US aluminum producers for testing smooth specimens of 7000 series alloys containing no more than 0.25% copper. The advantages of this test are indicated in the figure above.

This test is not effective for the 2000 series (Al-Cu) alloys, the 5000 series (Al-Mg) alloys, or the 7000 series (Al-Zn-Mg-Cu) alloys containing more than 1% copper. Also, it is not recommended for testing precracked specimens because of the interference of wedges of corrosion products developed on the crack surfaces.

Various Aluminum Alloys: Stress-Corrosion Cracking, in Sodium Chloride

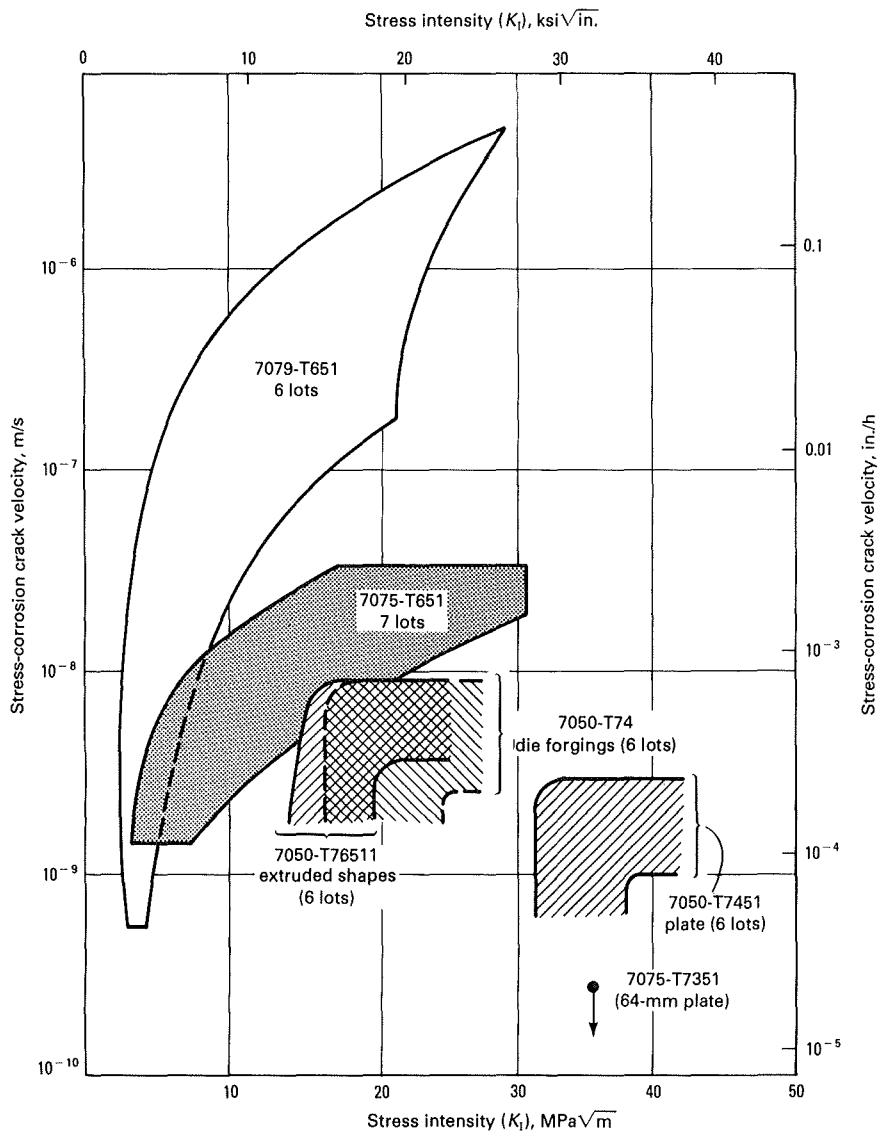


Stress-corrosion cracking in 3.5% sodium chloride solution for several aluminum alloys and heat treatments.



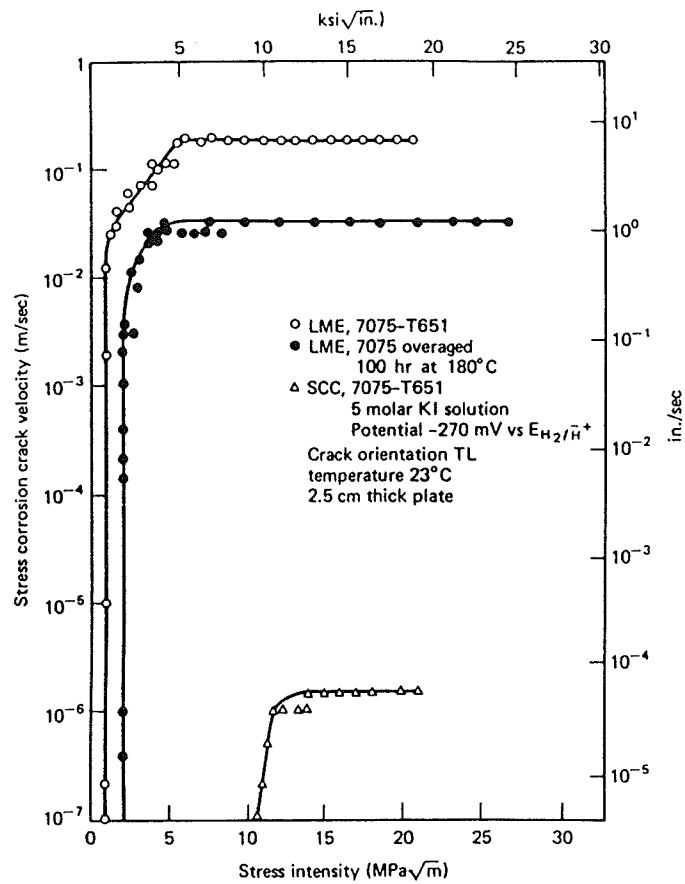
Temperature dependence on stress-corrosion cracking of
7039-T64 at applied stresses of 5 to 55 ksi.

7075 Aluminum Alloy: Stress-Corrosion Cracking Propagation Rates, in Sodium Chloride



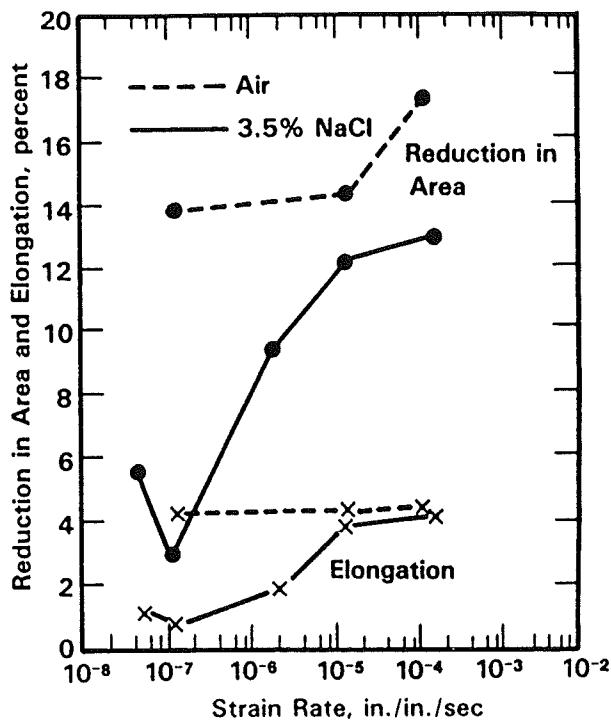
Stress-corrosion cracking propagation rates for various 7075 aluminum alloy products.

The specimens were double cantilever beams tested in the S-L plate orientation. They were bolt-loaded to pop-in and wetted three times daily with 3.5% sodium chloride. The right end of the band for each product indicates the pop-in starting stress-intensity.

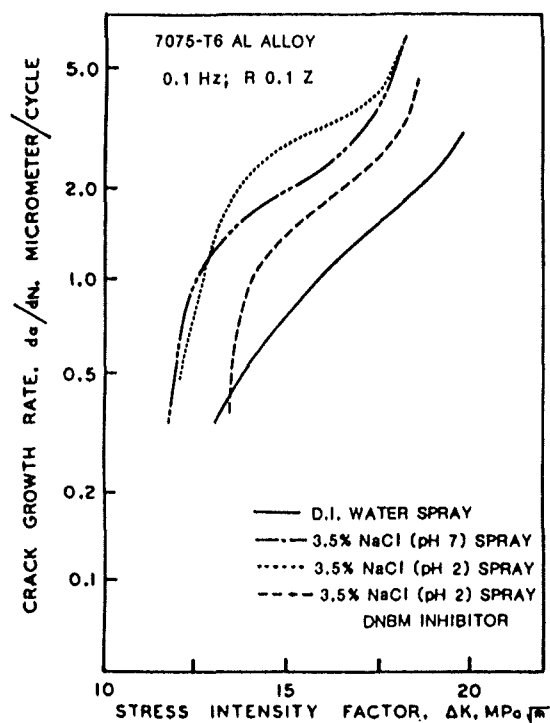


Environment-assisted cracking with liquid mercury and aqueous iodide solution in 7075 aluminum alloy.

7075-T6 Aluminum Alloy: Slow Strain Rate Stress-Corrosion Cracking Results

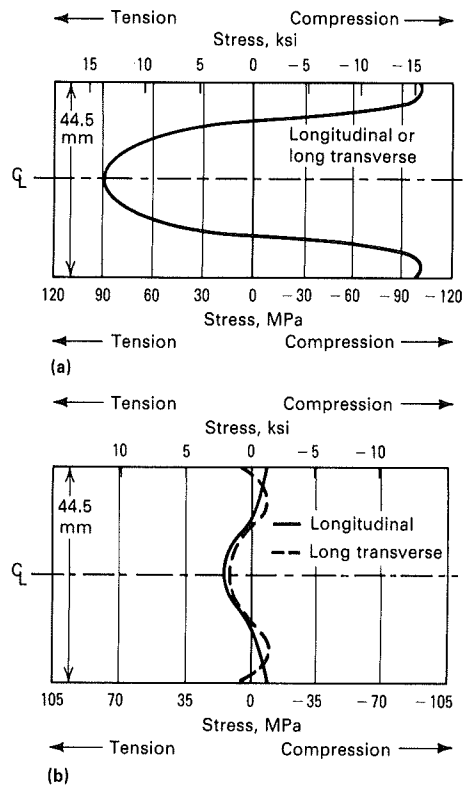


Slow strain rate stress-corrosion cracking test results for 7075-T6.



Effect of dichromate + nitrite + borate + molybdate on corrosion fatigue crack growth rate for aluminum alloy 7075-T6.

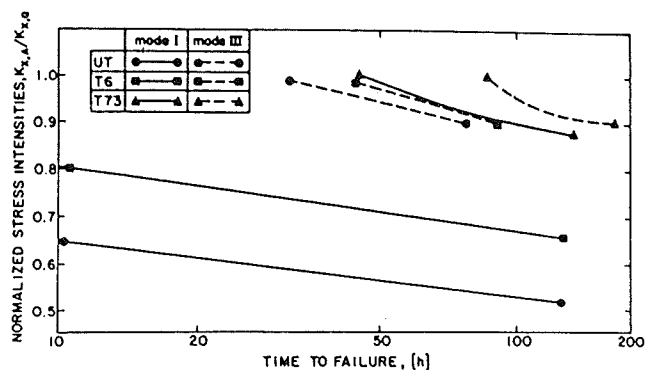
7075-T6 Aluminum Alloy Plate: Comparison of Residual Stresses Before and After Stress Relief, in Sodium Chloride Solution



Comparison of residual stresses in a thick, constant cross section 7075-T6 aluminum alloy plate before and after stress relief. (a) High residual stresses in the solution treated and quenched condition. (b) Reduction in stresses after stretching 2%.

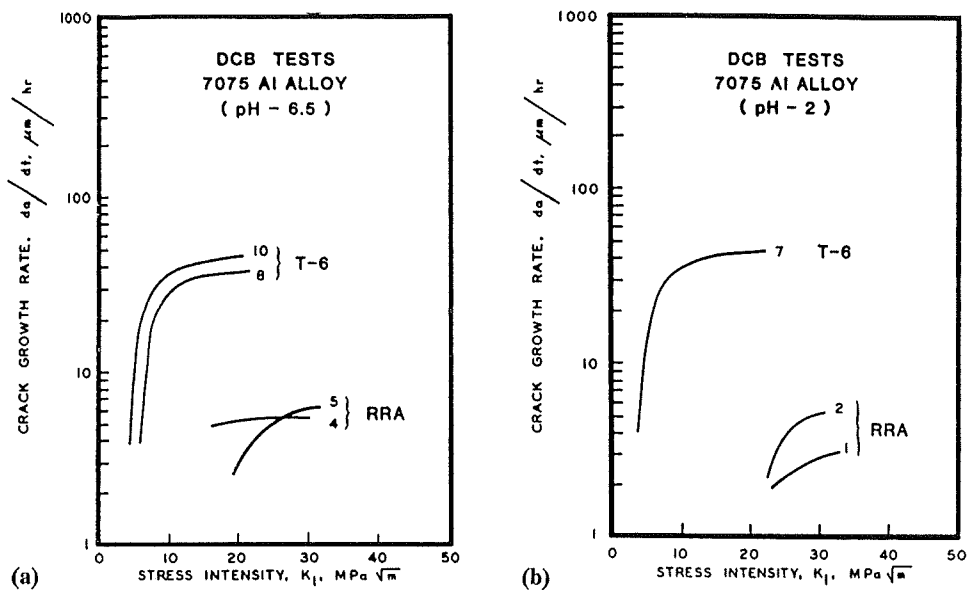
7075-UT, T6, AND T73 Aluminum Alloys: Summary of Stress-Corrosion Cracking Results in Aluminum Chloride

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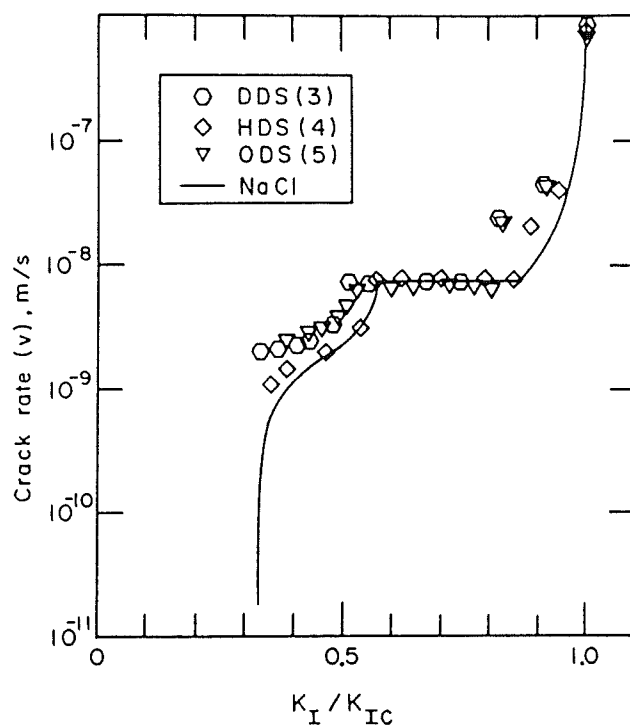


Summary of 1N aluminum chloride stress-corrosion cracking results for Modes I and III of 7075-UT, T6, and T73.

7075-T651 and RRA Aluminum Alloys: Stress-Corrosion Crack Growth Rate, in Sodium Chloride Solution

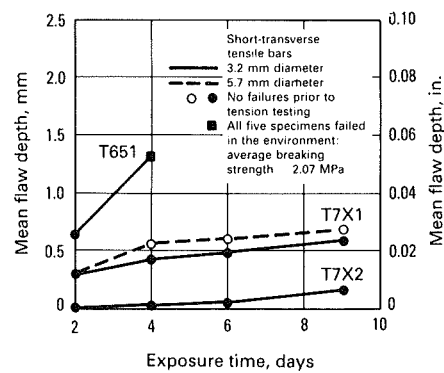


Stress-corrosion crack growth rate for 7075-T651 and RRA tested in 3.5% sodium chloride solution. (a) pH 6.5. (b) pH 2.



Rate of stress-corrosion crack growth as a function of K_I/K_{IC} for 7075-T651 alloy in 3.5% sodium chloride + sodium alkyl sulfonate (DDS, HDS, and ODS, with the dodecane-, hexadecane-, and octadecane- groups, respectively). The solid line represents behavior in 3.5% sodium chloride alone.

7075-T651 Aluminum Alloy: Effect of Temper on Stress-Corrosion Cracking Resistance, in Sodium Chloride Solution

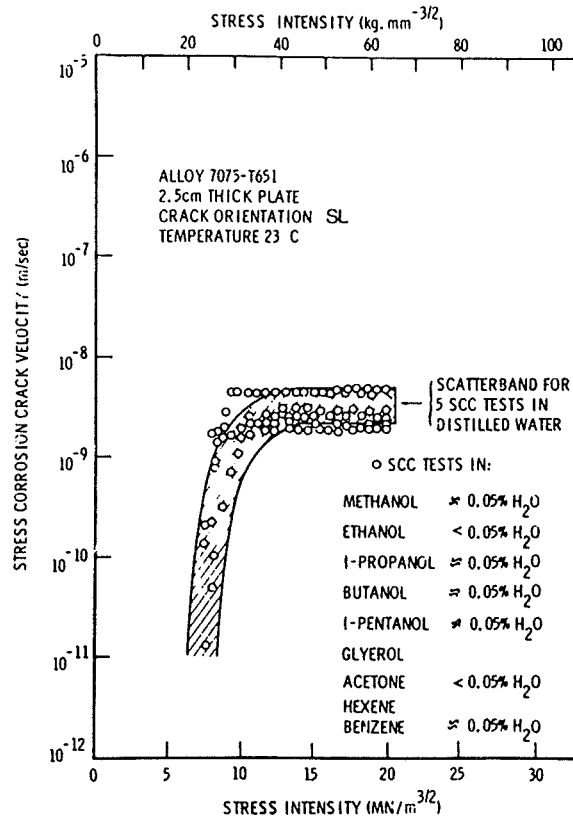


Effect of temper on stress-corrosion cracking resistance performance of aluminum alloy 7075 subjected to alternate immersion in 3.5% sodium chloride solution at a stress of 207 MPa (30 ksi).

By using an elastic-plastic fracture mechanics model, the mean flaw depth is calculated from the mean breaking stress, the strength, and the fracture toughness of the test material. The test results indicate that the thermal treatments used to reduce the stress-corrosion cracking susceptibility of 7075-T651 decreased stress-corrosion cracking penetration.

7075-T651 Aluminum Alloy: Effect of Various Organic Liquids on Subcritical Crack Growth

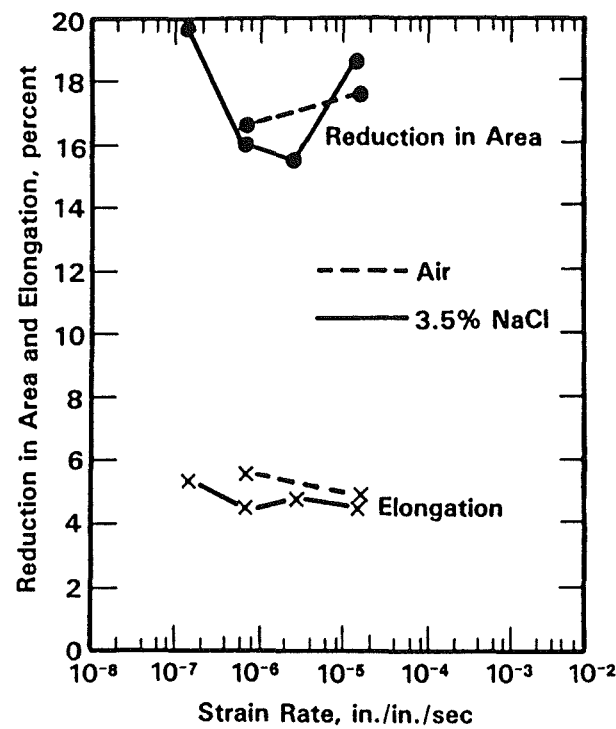
277



Effect of various organic liquids on subcritical crack growth in a high-strength aluminum alloy.

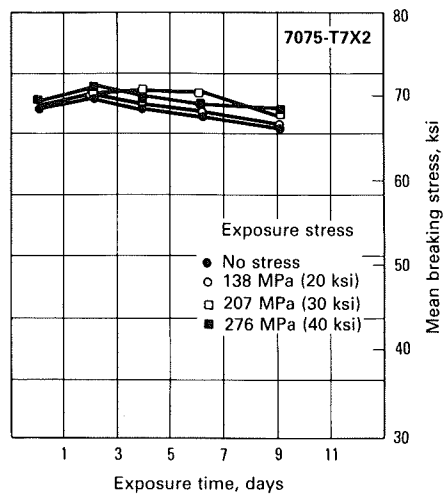
Note that the crack growth rates in all organic liquids containing traces of water fall into the scatterband for stress-corrosion cracking tests in pure water.

7075-T7351 Aluminum Alloy: Slow Strain Rate Stress-Corrosion Cracking Results



Slow strain rate stress-corrosion cracking test results for 7075-T7351.

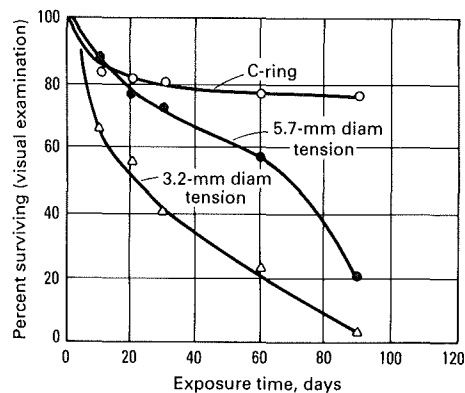
Source: J.A. Beavers, G.H. Koch, and W.E. Berry, Corrosion of Metals in Marine Environments, Metals and Ceramics Information Center, Battelle Columbus Laboratories, July 1986.



Mean breaking stress after exposure as a function of exposure time for short-transverse 3.2-mm diameter aluminum alloy 7075 tension specimens tested according to ASTM G 44 at various exposure stress levels. Each point represents an average of five specimens.

In studying trends in stress-corrosion cracking susceptibility, replicate sets of specimens stressed at several stress levels including zero are often used. This procedure is illustrated above for samples of 7075 that have been heat treated to decrease susceptibility to stress-corrosion cracking. Analysis of these breaking stress data by extreme value statistics enables calculation of survival probabilities and the estimation of a threshold stress, without depending on failures during exposure.

7075-T7X51 Aluminum Alloy: Influence of Specimen Configuration on Stress-Corrosion Cracking, in Sodium Chloride Solution

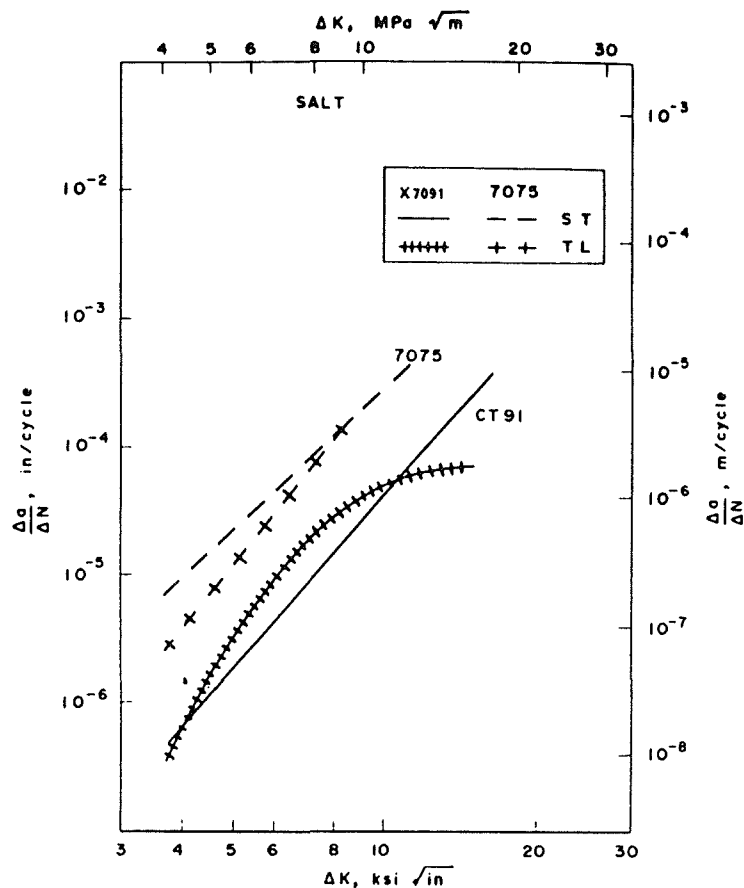


Influence of specimen configuration on stress-corrosion cracking test performance (alternate immersion in 3.5% sodium chloride per ASTM G 44). Aluminum alloy 7075-T7X51 specimens were stressed at 310 MPa (45 ksi). Each point represents 60 to 90 specimens.

This figure shows the influence of the type of specimen used on stress-corrosion cracking ratings obtained in traditional pass-fail methods. An advantage of using the mean flaw depth to evaluate stress-corrosion cracking performance is that the effects of specimen size and alloy strength and toughness can be normalized.

7075-T651 and P/M X7091-T7E70 Aluminum Alloys: Corrosion Fatigue Behavior, in Sodium Chloride Solution

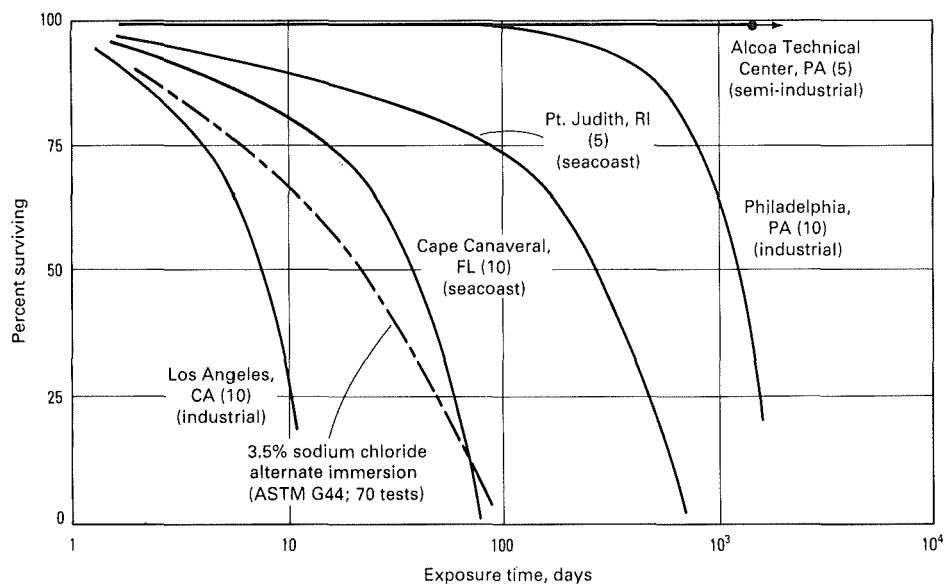
281



Corrosion fatigue behavior of 7075 and X7091 (CT91) alloys in 3.5% sodium chloride solution.

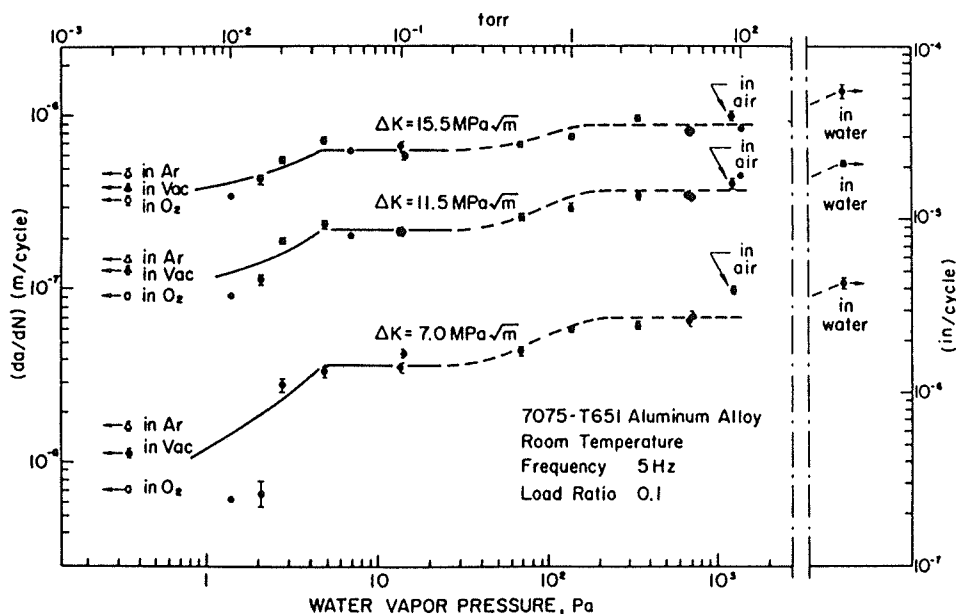
The 7075 alloy was in the T651 temper, and the X7091 alloy, a powder metallurgy product, was in the T7E70 temper. The test frequency was 0.2 Hz. The presence of a measurable crack closure level for P/M X7091 correlated with a lower corrosion fatigue crack propagation rate. The formation of corrosion products on the crack surfaces increased the closure levels.

7075-T7651 Aluminum Alloy Plate: Effect of Variations in Atmospheric Environment



Effect of variations in atmospheric environment on the probability and time-to-failure by stress-corrosion cracking of a material with an intermediate susceptibility. Tests were made on short-transverse 3.2-mm (0.125-in.) diameter tension specimens from 7075-T7651 plate stressed at 310 MPa (45 ksi). Parenthetical values indicate the number of tests involved.

The relatively conservative estimate of the stress-corrosion cracking resistance of an intermediate resistance alloy is shown, as is the wide range in behaviors in various atmospheric environments.

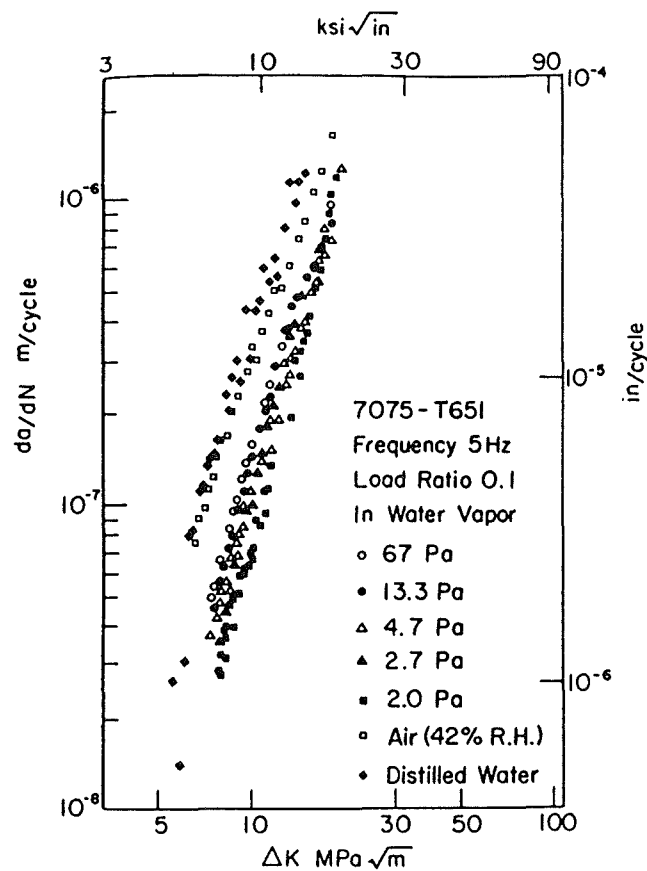


Influence of water vapor pressure on the fatigue crack growth rate in I/M 7075-T651 aluminum alloy at room temperature and $R = 0.1$.

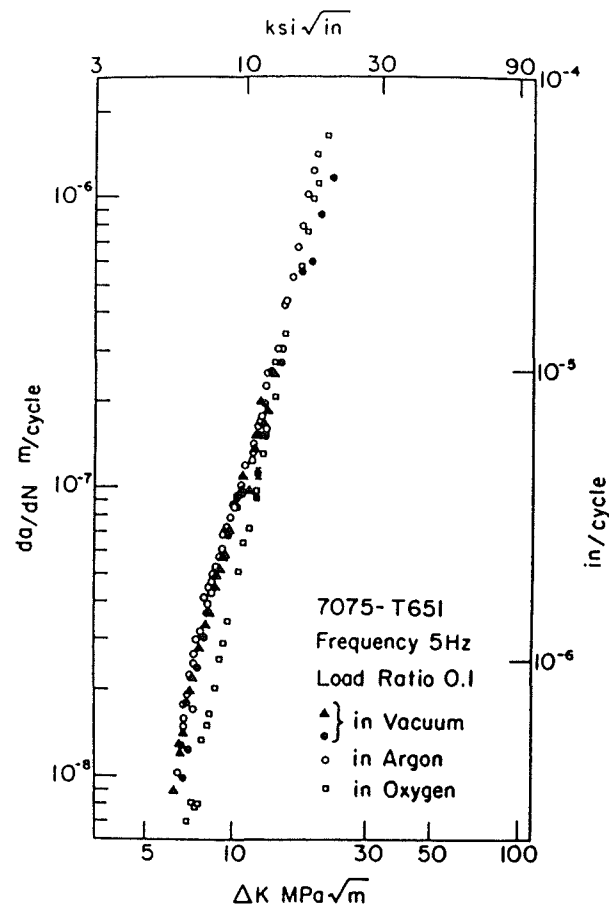
In the low-pressure region, fatigue crack growth rate increased with water vapor pressure up to about 4.7 Pa and then became essentially independent of pressure up to about 67 Pa. Above 67 Pa, additional increases in growth rate occurred with further increases in pressure. In the low-pressure region, response is controlled by the rate of transport of water vapor to the crack tip by molecular flow. In the high-pressure region, it appeared that response is surface reaction controlled, involving reactions of water vapor with segregated aluminum.

Fractographic analysis indicated that corrosion fatigue was crystallographic in nature, with propagation occurring along $\{100\}$ planes, whereas the pure fatigue process was noncrystallographic.

I/M 7075-T651 Aluminum Alloy: Effect of Water Vapor Pressure on Fatigue Crack Growth



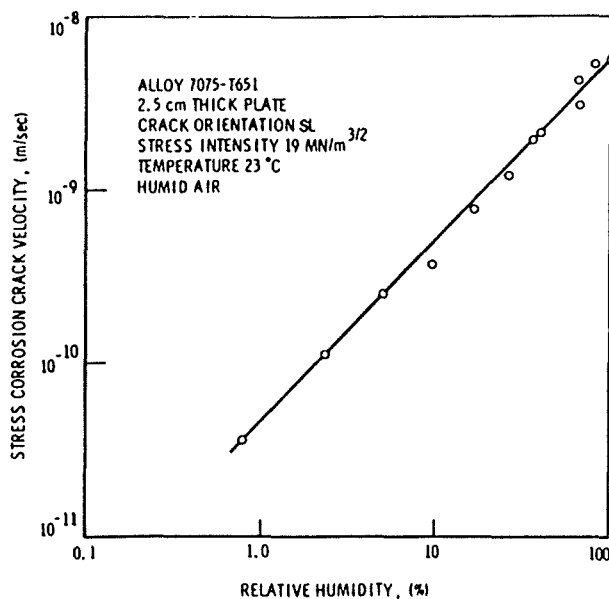
Effect of water vapor pressure on fatigue crack growth kinetics for I/M 7075-T651.



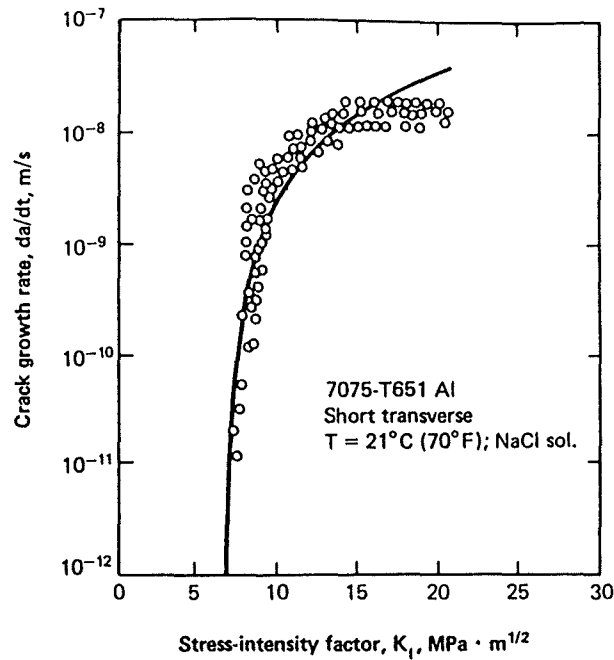
Comparison of the kinetics of fatigue crack growth in I/M 7075-T651 in vacuum, pure argon, and in pure oxygen.

The reasons for the faster growth rates in vacuum and argon relative to pure oxygen are not understood.

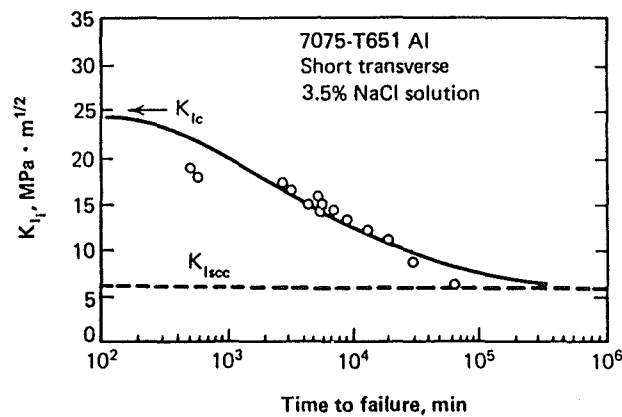
7075-T651 Aluminum Alloy: Effect of Humidity on Stress-Independent Subcritical Crack Growth



Effect of humidity on the stress-independent subcritical crack growth rate in a high-strength aluminum alloy exposed to humid air.



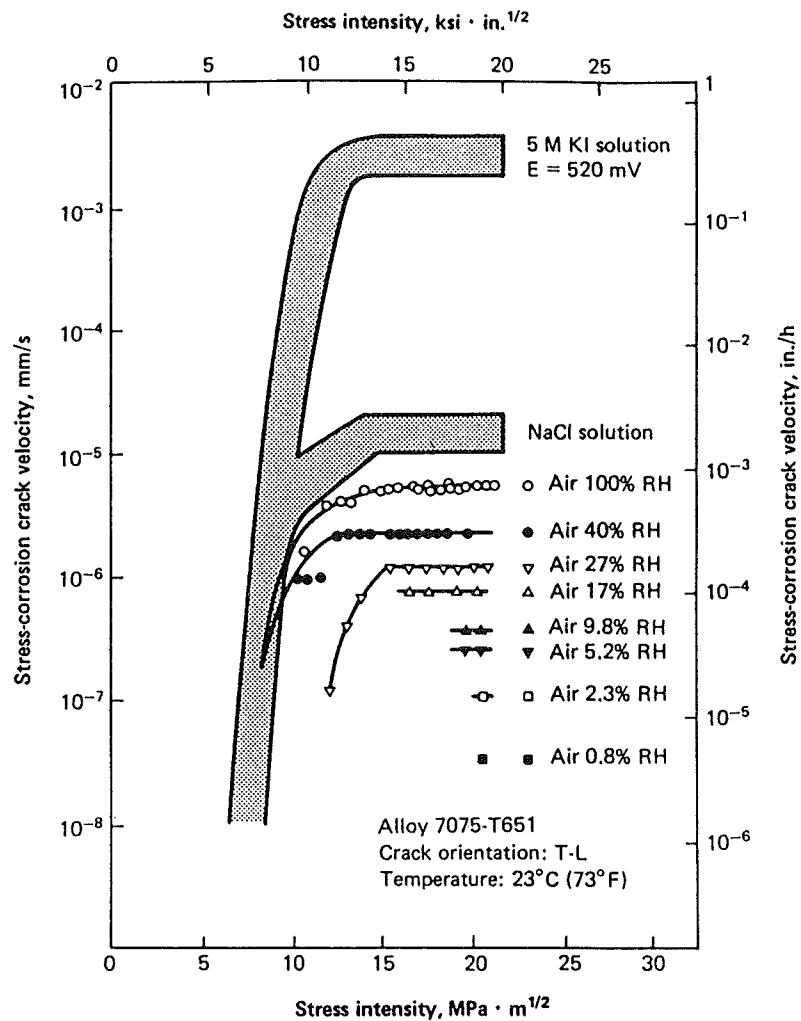
(a)



(b)

Results of room-temperature stress-corrosion tests on pre-cracked specimens of aluminum alloy 7075-T651.

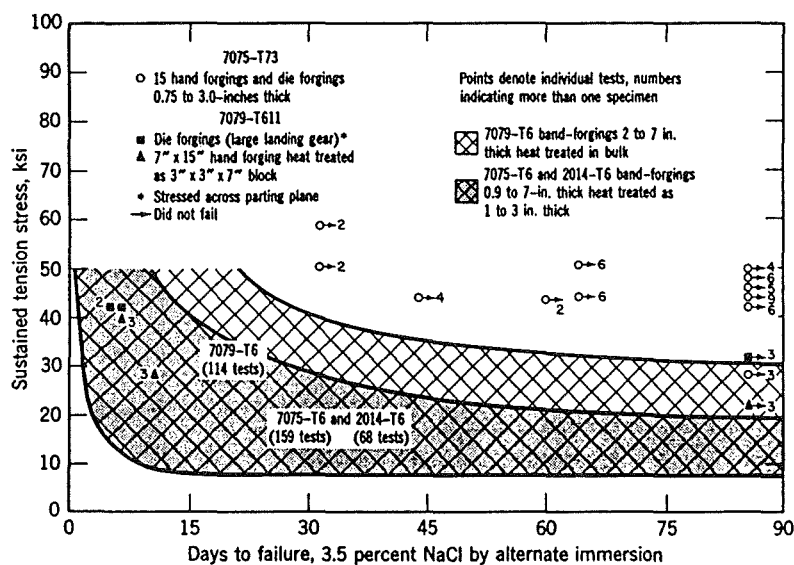
7075-T651 Aluminum Alloy: Effect of Relative Humidity and Stress-Intensity Factor on Stress-Corrosion Crack Velocity



Effect of relative humidity and stress-intensity factor on the stress-corrosion crack velocity in a high-strength aluminum alloy.

7075-T73 and 7079-T611 Aluminum Alloys: Relative Resistance to Stress-Corrosion Cracking

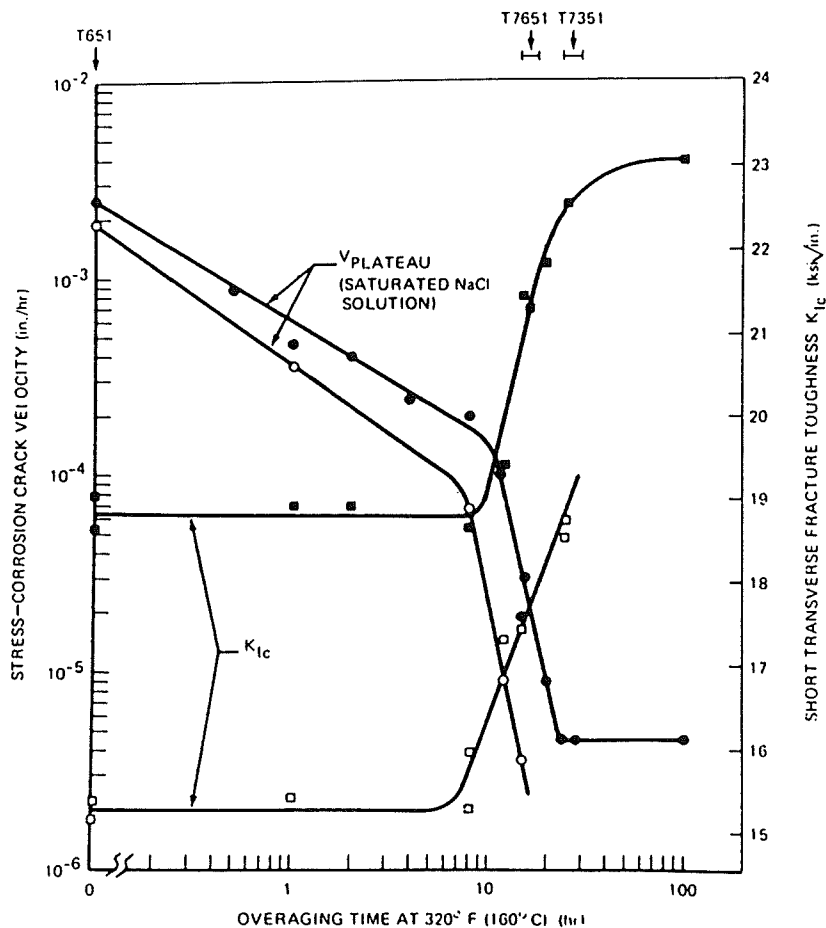
289



Relative resistance to stress-corrosion cracking of high-strength aluminum alloy forgings tested in the short-transverse direction.

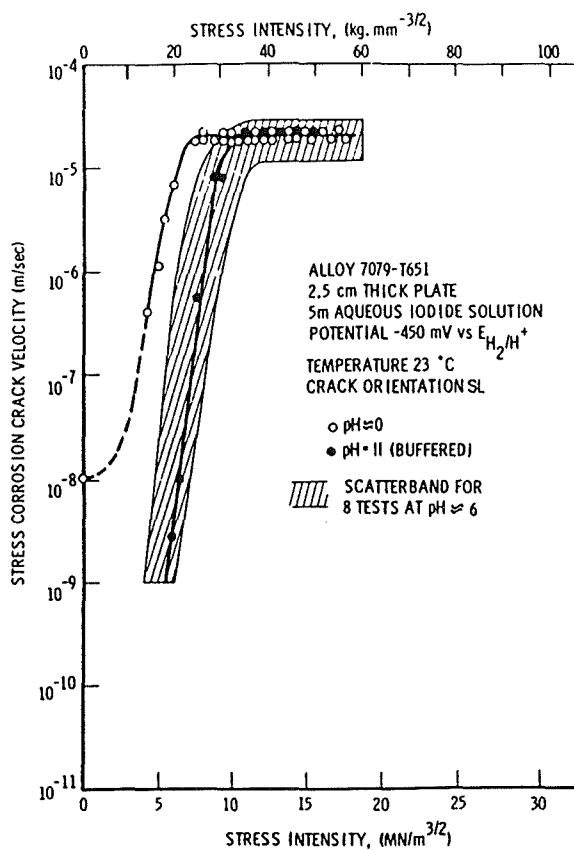
Specific test data for 7075-T73 and 7079-T611 are plotted on the background of test data for the established materials 2014-T6 and 7075-T6.

7075-T651 and 7178-T651 Aluminum Alloys: Effect of Aging Time on Short-Transverse Toughness and Stress-Corrosion Crack Velocity



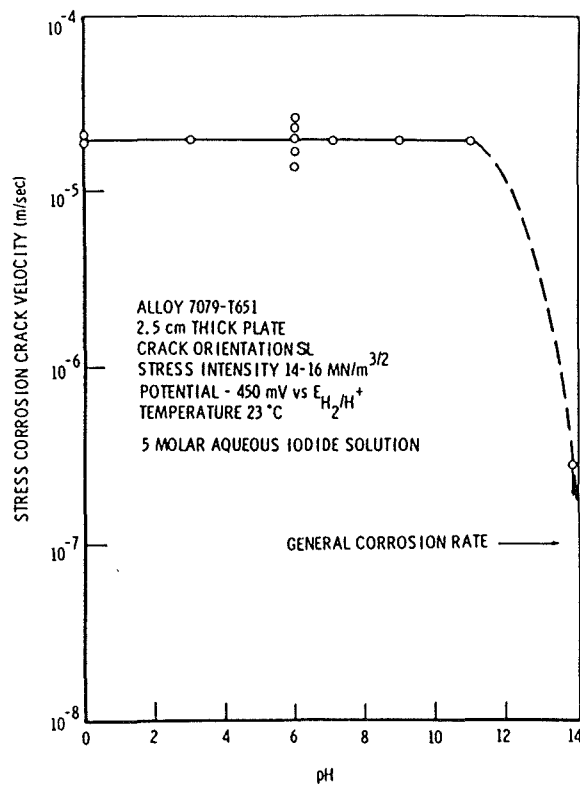
Effect of aging time at 160 °C (320 °F) on the short-transverse toughness and stress-corrosion crack velocity of 7075-T651 (closed symbols) and 7178-T651 (open symbols). The duration of the overaging treatment for various tempers is indicated at the top of the figure.

Overaging up to 10 h can reduce the stress-corrosion cracking velocity by an order of magnitude without affecting the fracture toughness (or strength levels).

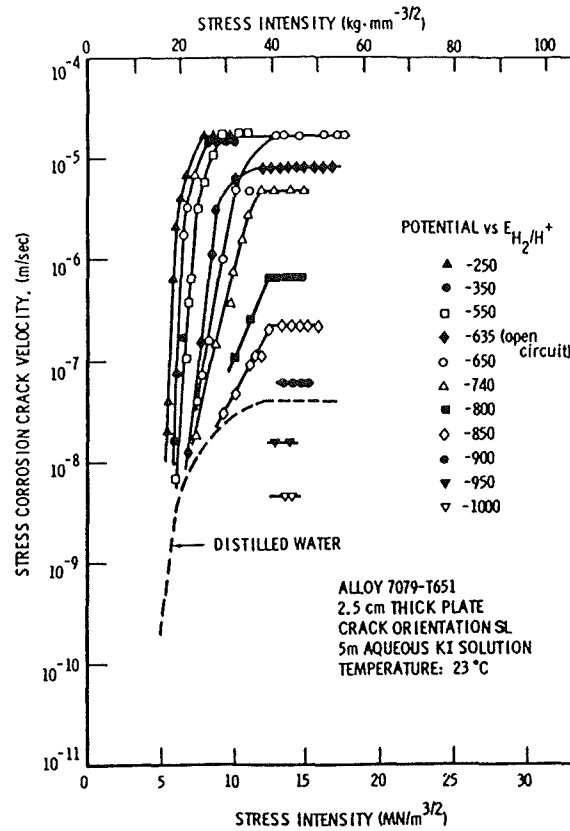


Effect of stress-intensity and pH on the velocity of stress-corrosion cracks in a high-strength aluminum alloy.

7079-T651 Aluminum Alloy: Effect of pH on the Plateau Growth Rate of Stress-Corrosion Cracks

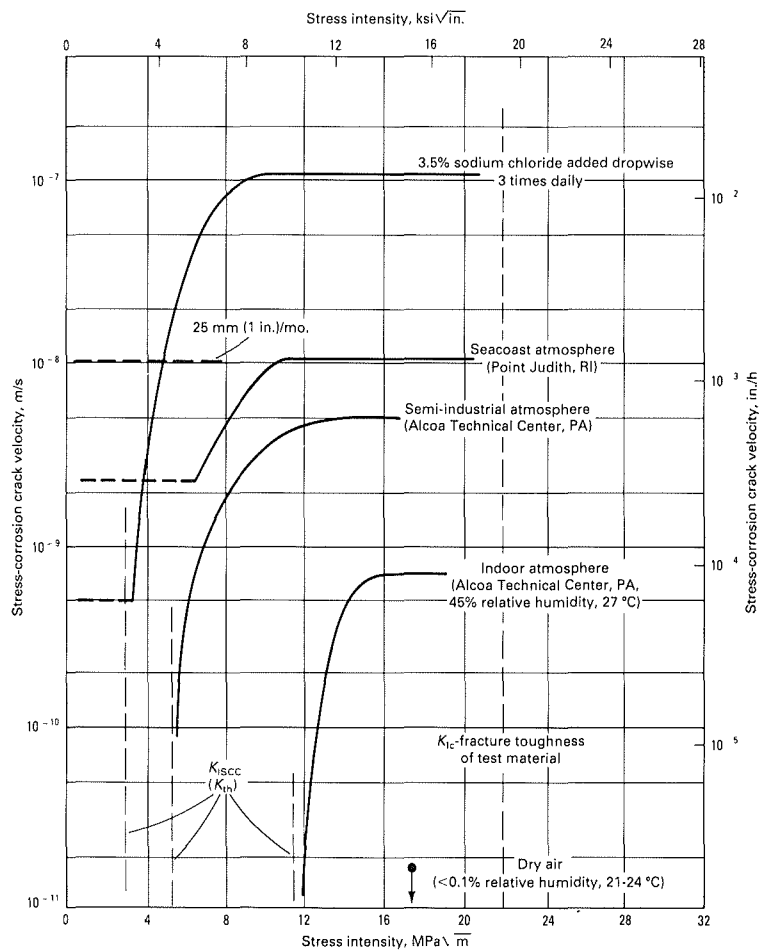


Effect of pH on the stress-independent (plateau) growth rate of stress-corrosion cracks in a high-strength aluminum alloy.



Effect of electrode potential and stress-intensity on stress-corrosion crack velocity in a high-strength aluminum alloy.

7079-T651 Aluminum Alloy Plate: Effects of Corrosive Environment

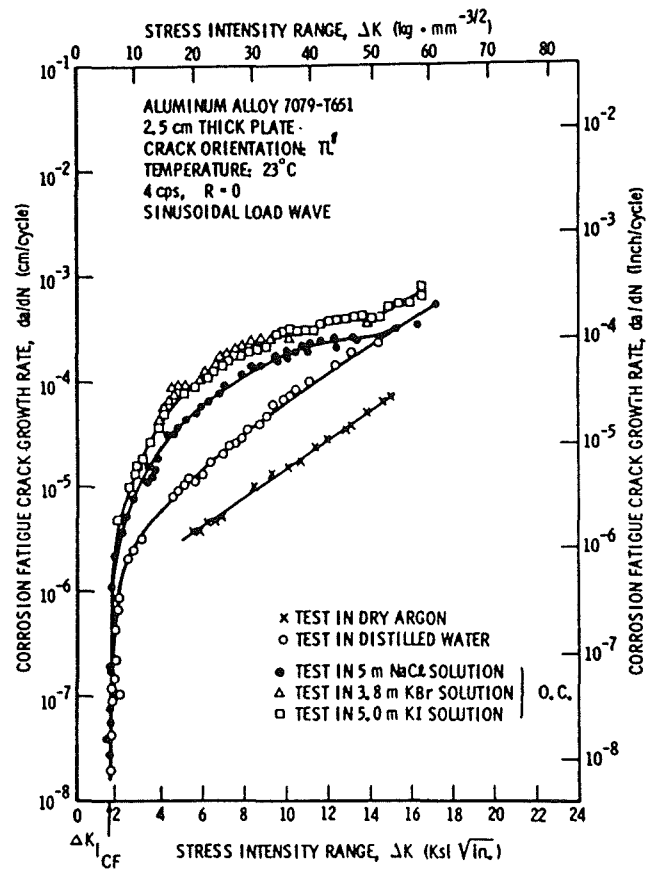


Effect of corrosive environment on stress-corrosion cracking velocity and threshold stress-intensity for 64-mm thick 7079-T651 plate stressed in the S-L orientation.

The specimens were double cantilever beams bolt-loaded to pop-in. No stress-corrosion cracking occurred during 3 years of exposure to dry air. The plateau velocity and the apparent threshold intensity (K_{ISCC} or K_{th}) vary with the environment. It is interesting to note that corrosion product wedging occurred in the regions denoted by dashed lines. This wedging props open the crack and prevents any further decrease in K-level with an increase in crack length.

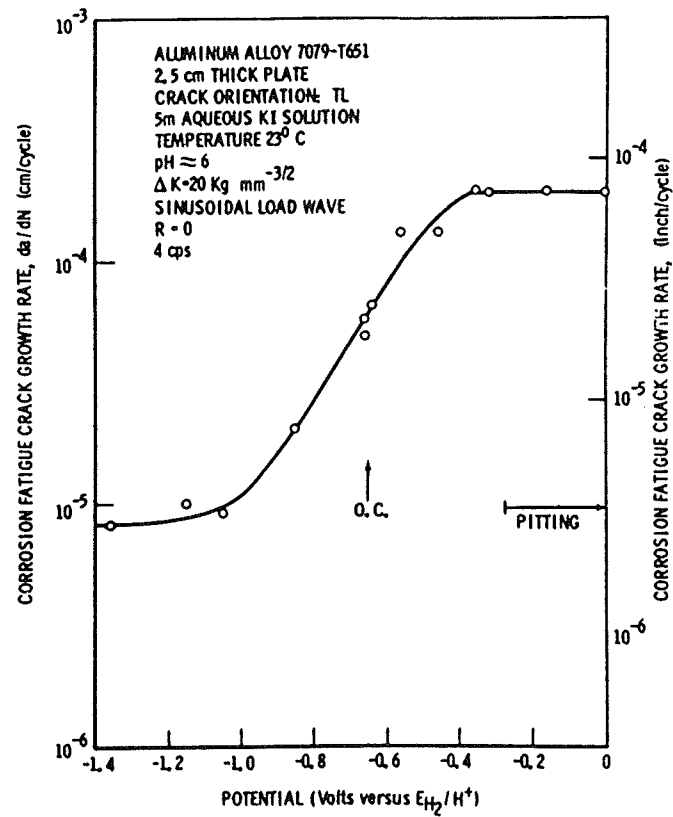
7079-T651 Aluminum Alloy: Effect of Cyclic Stress-Intensity Range on Growth of Corrosion-Fatigue Cracks

295



Effect of cyclic stress-intensity range on the growth rate of corrosion-fatigue cracks in a high-strength aluminum alloy exposed to various environments.

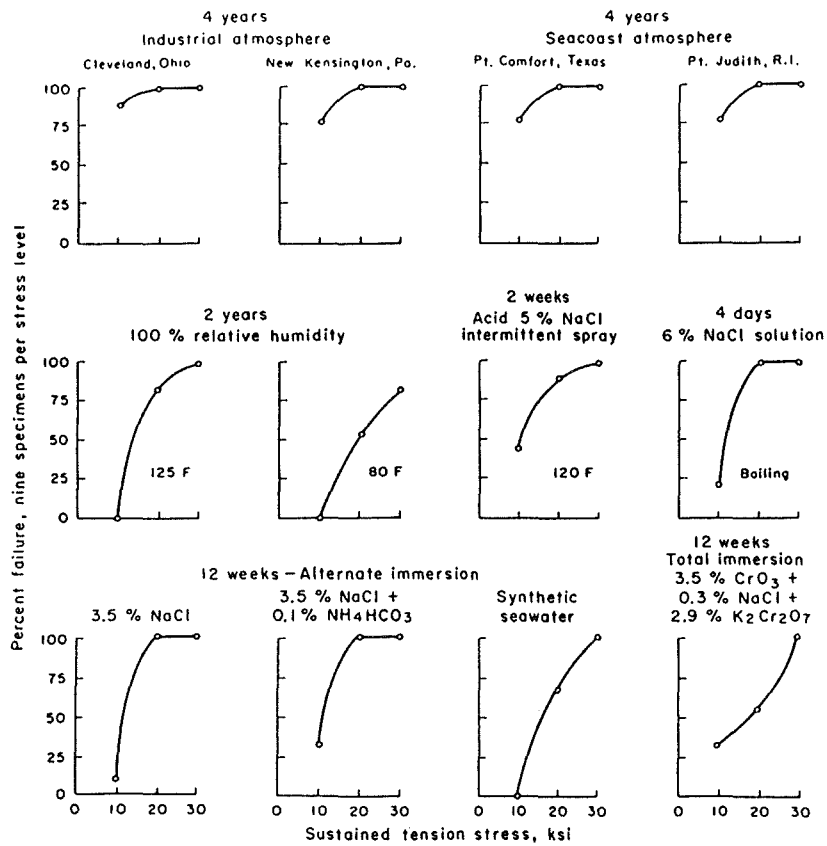
7079-T651 Aluminum Alloy Plate: Effect of Potential on Growth Rate of Corrosion-Fatigue Cracks, in an Aqueous Halide Solution



Effect of potential on the growth rate of corrosion-fatigue cracks in a high-strength aluminum alloy exposed to an aqueous halide solution (O.C. = open circuit potential).

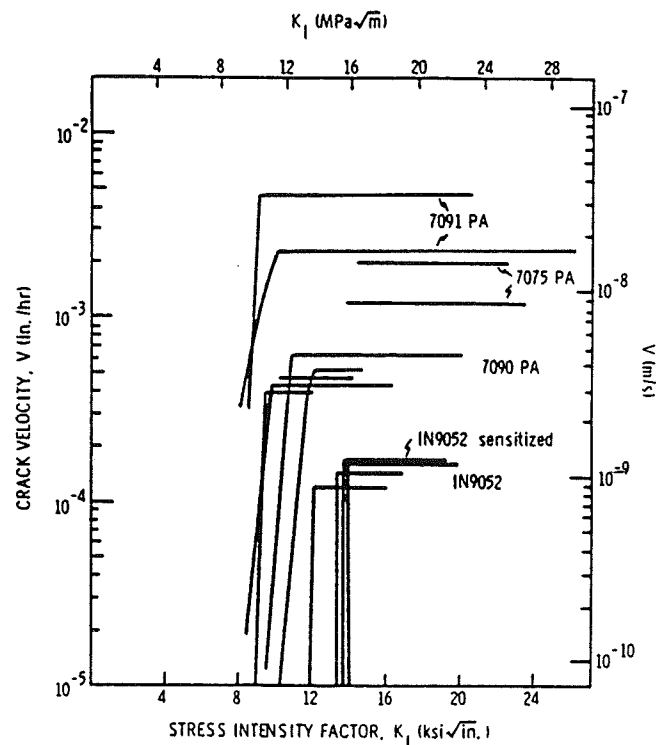
7090-T6 Aluminum Alloy Die Forgings: Effect of Environment on Resistance to Stress-Corrosion Cracking

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Effect of environment on resistance to stress-corrosion cracking of 7090-T6 die forgings stressed in the short-transverse direction.

**7090-T6511, 7091-T6E192, and IN9052 F Temper P/M
Aluminum Alloys and I/M Aluminum Alloy 7075-T6511:
Stress-Corrosion Crack Velocity as a Function of
Stress-Intensity Factor**



Stress-corrosion crack velocity as a function of stress-intensity factor (v - K) curves for three P/M alloys and I/M 7075, each in its highest strength condition.

The specimens were tested as bolt-loaded double cantilever beams in a chromate-inhibited brine solution. (PA = peak strength condition). Alloy 7075 is an ingot metallurgy (I/M) product. Alloys 7090, 7091, and IN9052 are rapidly solidified powder metallurgy (RS-P/M) products.

Composition of alloys (wt%):

7075: 6.0 Zn, 2.3 Mg, 1.62 Cu, 0.26 Fe, 0.12 Si, 0.21 Cr, bal Al

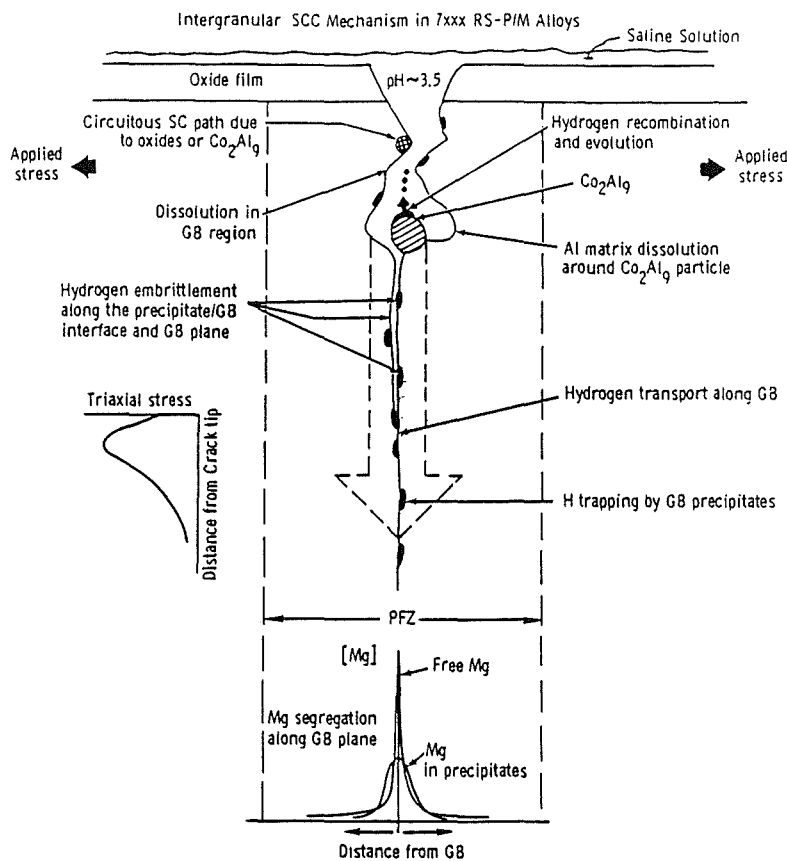
7090: 7.7 Zn, 2.3 Mg, 1.00 Cu, 1.65 Co, bal Al

7091: 6.5 Zn, 2.4 Mg, 1.55 Cu, 0.44 Co, bal Al

IN9052: 3.94 Mg, 0.33 O, 0.91 C, bal Al

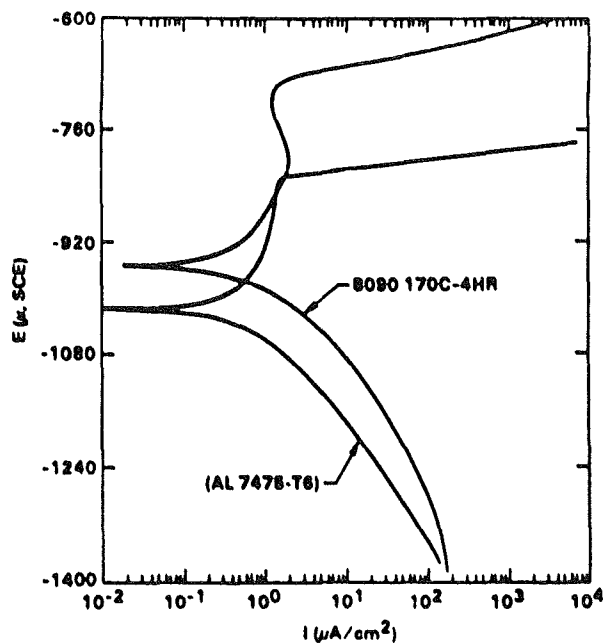
**7090-T6511, 7091-T6E192, and IN9052 F Temper P/M
Aluminum Alloys and I/M Aluminum Alloy 7075-T6511:
Stress-Corrosion Cracking Mechanism**

299

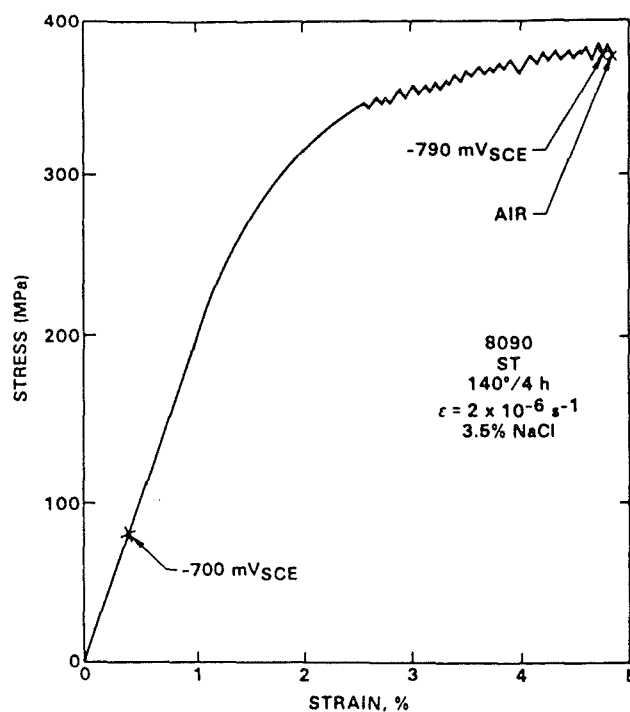


Schematic of stress-corrosion cracking mechanism in cobalt-containing RS-P/M alloys (PFZ = precipitate-free zone).

8090 and 7475-T6 Aluminum Alloys: Potentiodynamic Polarization Curves, in Sodium Chloride Solution

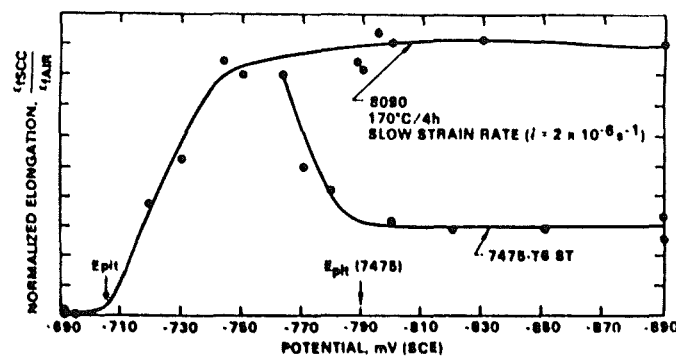


Potentiodynamic polarization curves for an Al-9Li alloy, 8090 alloy, aged at 170 °C for 4 h and an Al-Zn-Mg-Cu alloy, AA 7475-T6, in 0.5M sodium chloride solution.



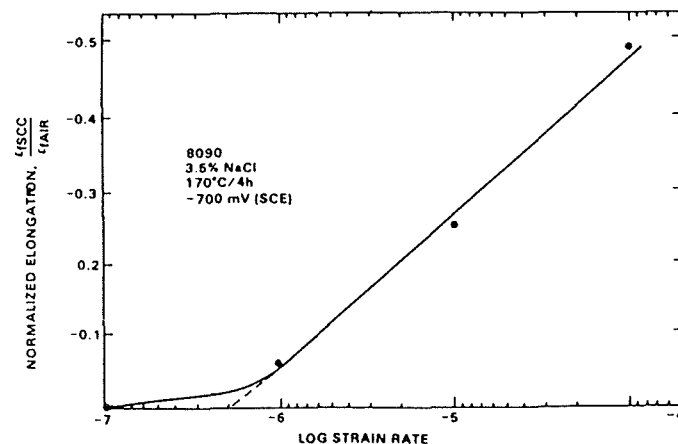
Slow stress-strain curve in air of Alloy 8090 aged at 170 °C for 4 h with points showing where fracture occurred in 0.5M sodium chloride at -700 and -790 mV_{SCE}.

8090 and 7475-T6 Aluminum Alloys: Effects of Potential on the Ratio of the Strain-to-Failure in Sodium Chloride to that in Air



Effects of potential on the ratio of the strain-to-failure in 0.5M sodium chloride solution to that in air for Alloy 8090 aged at 170 °C for 4 h and 7475-T6.

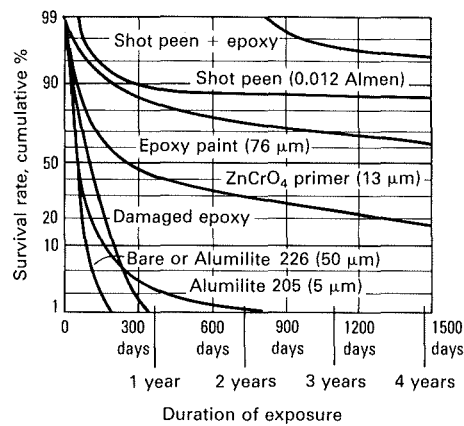
Below a potential of approximately -750 mV_{SCE} at strain rates on the order of 10⁻⁶/s, Alloy 8090 is highly resistant to stress-corrosion cracking. At potentials above -750 mV_{SCE}, the strain-to-failure ratio decreases rapidly with an increase in potential to the pitting potential. It was concluded that the critical potential controls the crack propagation rate and not the initiation rate.



Ratio of strain-to-failure in 0.5M sodium chloride (-700 mV_{SCE}) to the strain-to-failure in air for Alloy 8090 aged at 170 °C for 4 h at various strain rates.

2014-T651 and 7079-T651 Aluminum Alloys: Relative Effectiveness of Protective Systems in the Prevention of Stress-Corrosion Cracking

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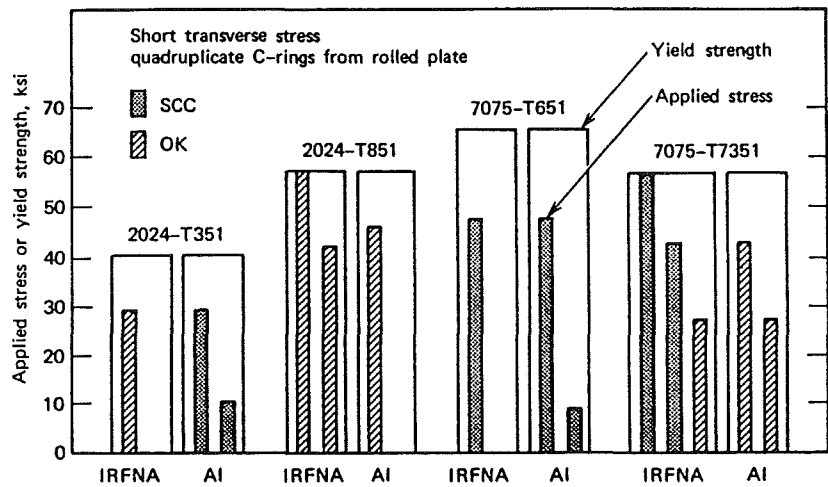


Relative effectiveness of various protective systems in preventing stress-corrosion cracking of susceptible aluminum alloys.

The curves represent combined results for specimens of 2014-T651 and 7079-T651 that had been highly strained in the elastic range. The specimens were exposed at Point Judith, Rhode Island; Comfort, Texas; and New Kensington, Pennsylvania. The beneficial effects of the residual compressive stresses introduced by shot peening are evident.

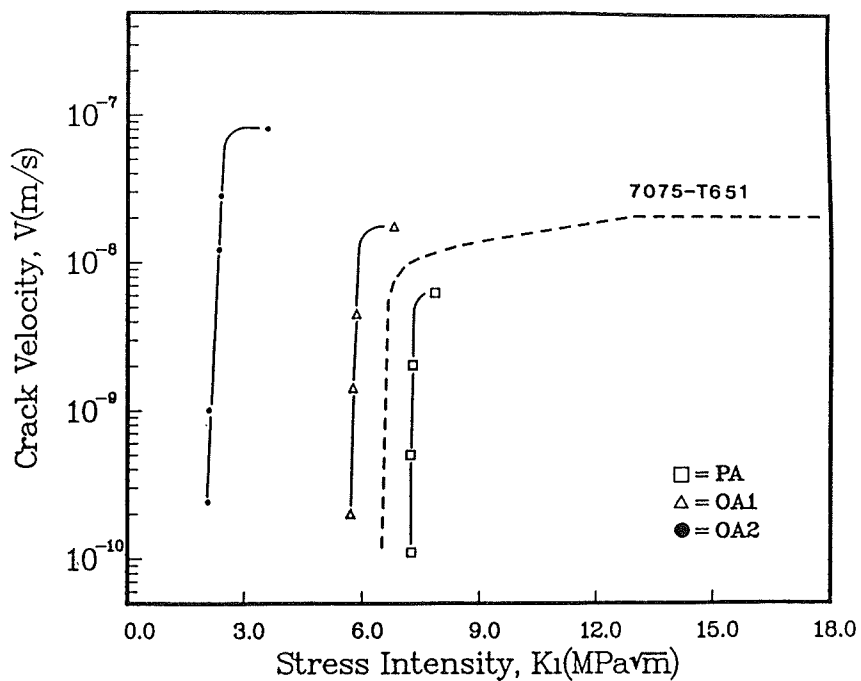
Residual stresses are introduced in aluminum alloy products when they are solution treated and quenched. Quenching places the surfaces in compression and the interior in tension. If no machining or stretching were done after quenching, the product would have enhanced resistance to stress-corrosion cracking. However, if a notch were to be machined into the region of tensile residual stress, then the stress-corrosion cracking resistance would be diminished. To avoid such a possibility, aluminum products of constant cross-sectional area are stress relieved by stretching after quenching to reduce the residual stresses.

2024-T351, 2024-T851, 7075-T651, and 7075-T7351 Aluminum Alloys: Comparison of Stress-Corrosion Cracking Resistance, in Nitric Acid and in Sodium Chloride



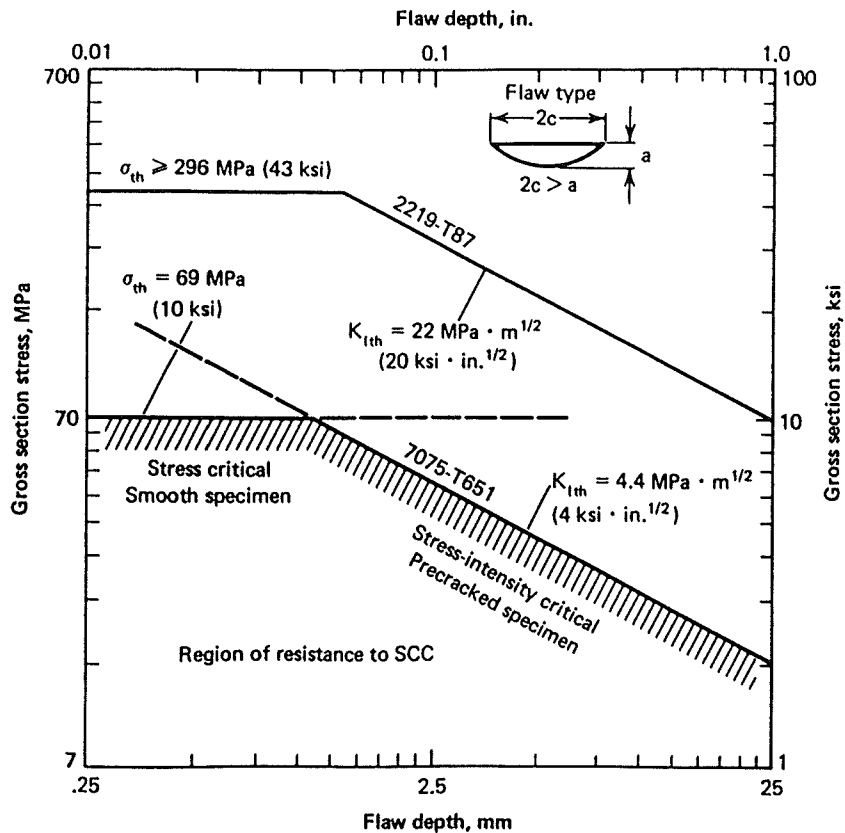
Comparison of resistance to stress-corrosion cracking of various alloys in inhibited red fuming nitric acid at 74 °C and in 3.5% sodium chloride alternate immersion.

Source: H.L. Craig, Jr., D.O. Sprowls, and D.E. Piper, Stress-Corrosion Cracking, in Handbook on Corrosion Testing and Evaluation, W.H. Ailor, Ed., John Wiley & Sons, New York, 1971, p 231-290.



Stress-corrosion crack growth rate versus stress-intensity for three tempers of 2090 and 7075-T651 alloy.

2219-T87 and 7075-T651 Aluminum Alloys: Stress and Stress-Intensity Characterizations, in a Salt-Dichromate-Acetate Solution

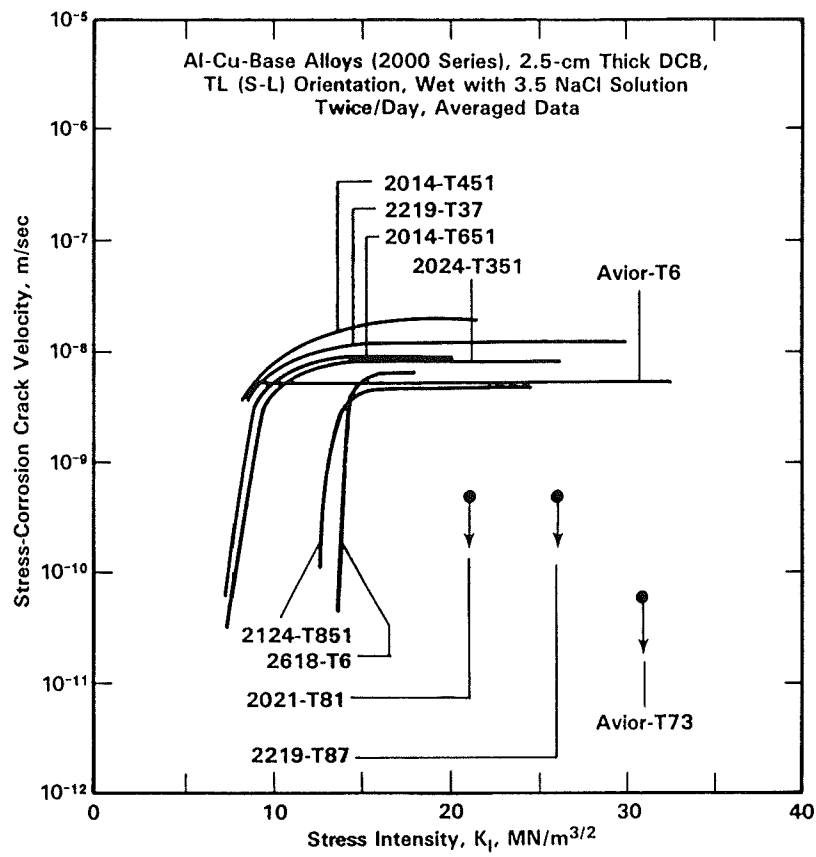


Stress and stress-intensity characterizations for two aluminum alloys exposed in salt-dichromate-acetate solution.

Note that at small crack sizes the applied stress rather than the applied stress-intensity becomes the controlling parameter. A similar transition occurs in the case of fatigue crack growth, where at small crack sizes the endurance limit becomes the controlling parameter for fatigue crack growth rather than the threshold stress-intensity level.

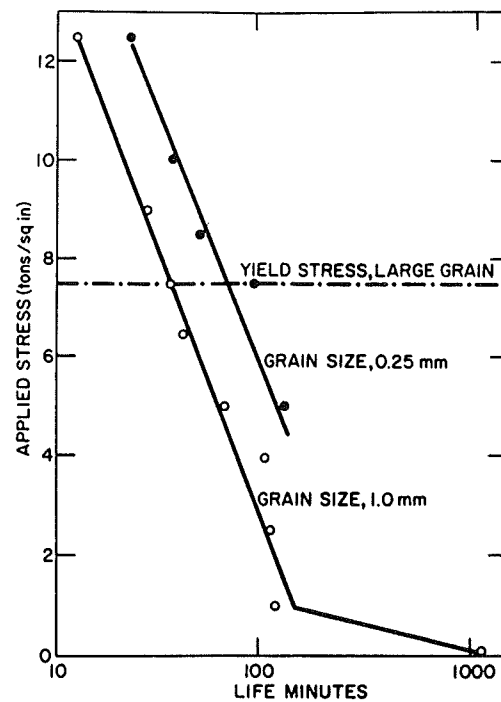
Al-Cu Alloys: Summary of Stress-Corrosion Cracking Growth Rates

307



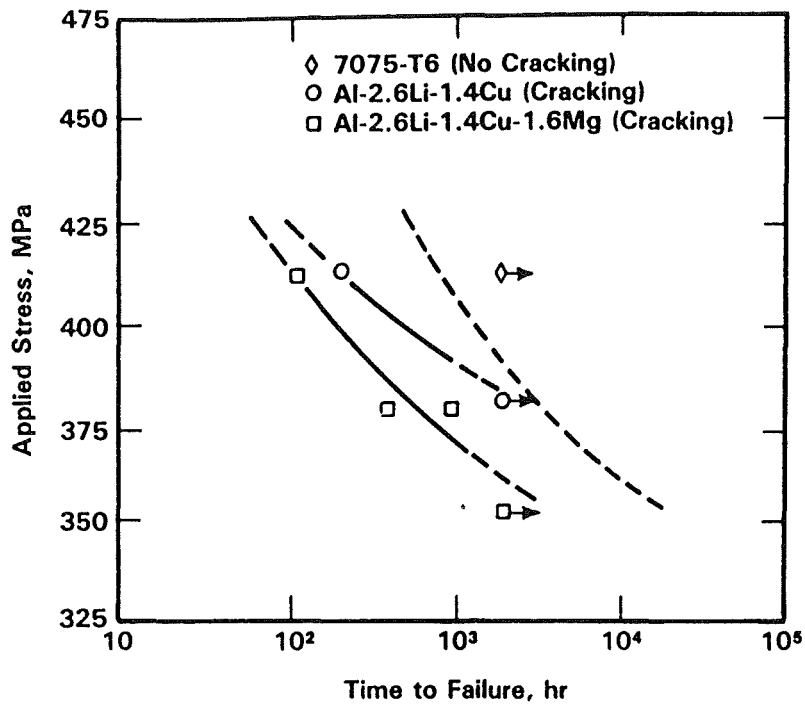
Summary of stress-corrosion cracking growth rates in various aluminum alloys based on the aluminum-copper system.

5Cu-Al Copper Alloy: Relationship of Applied Stress to Time-to-Fracture, in NaCl + NaHCO₃



Relationship of applied stress to time-to-fracture of 5Cu-Al alloy aged 1 h at 175 °C and exposed to 0.5N NaCl + 0.005N NaHCO₃.

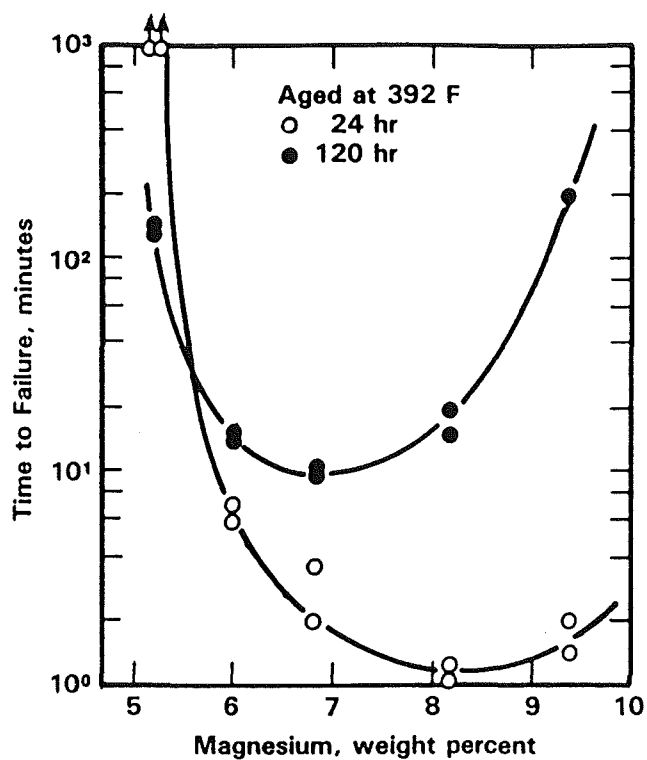
Al-Li-Cu P/M Alloys: Stress-Corrosion
Lifetime as a Function of Applied Stress, in
Sodium Chloride Solution



Stress-corrosion lifetime as a function of applied stress for two Al-Li-Cu powder metallurgy alloys tested in 3.5% sodium chloride solution as a function of applied stress.

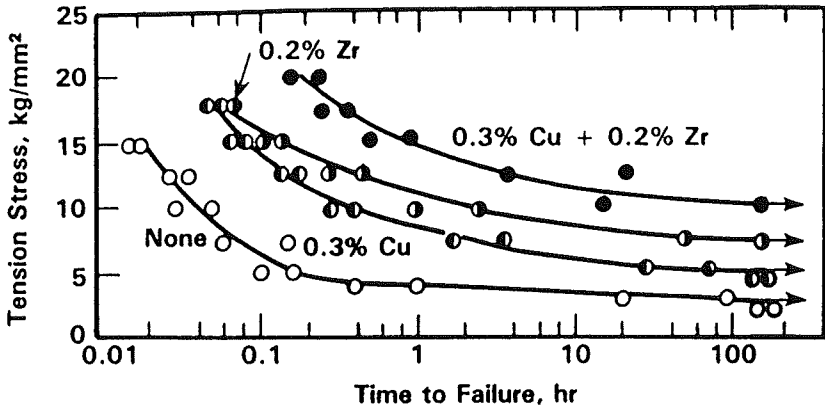
Source: J.A. Beavers, G.H. Koch, and W.E. Berry, Corrosion of Metals in Marine Environments, Metals and Ceramics Information Center, Battelle Columbus Laboratories, July 1986.

Al-Mg Alloys: Effect of Magnesium Content on Stress-Corrosion Cracking, in Sodium Chloride

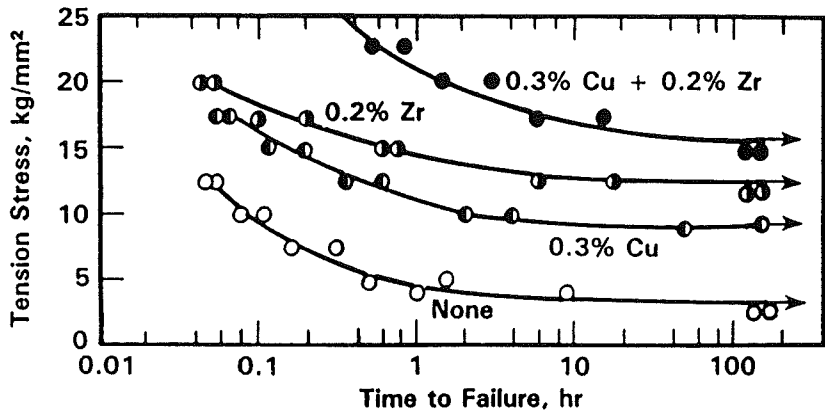


Effect of magnesium content on stress-corrosion cracking in 3.5% sodium chloride solution on Al-Mg alloys aged for either 24 or 120 h at 392 °F.

Al-8Mg Aluminum Alloy: Effects of Copper and Zirconium on Stress-Corrosion Cracking, in Sodium Chloride



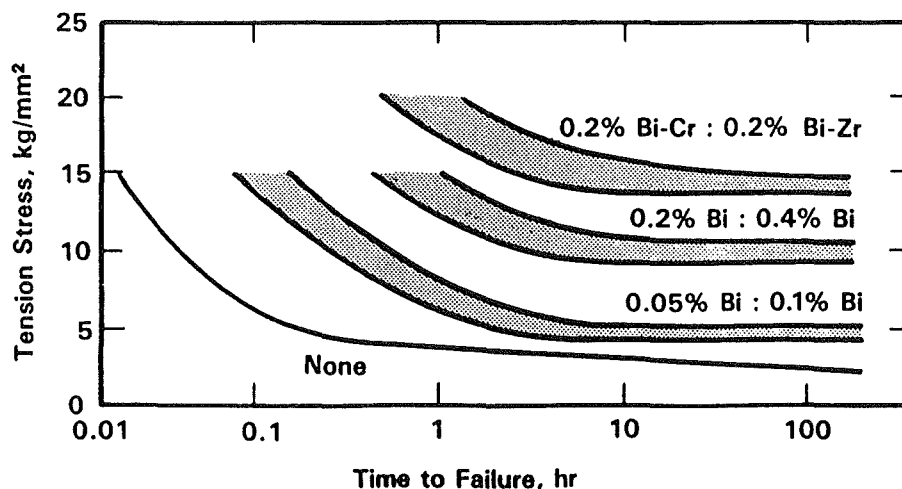
a. Water quenched from 750 F and aged at 265 F for 7 days



b. Air-Cooled from 750 F and aged at 265 F for 7 days

Effects of copper and zirconium on the stress-corrosion cracking of an Al-8Mg alloy tested under constant load in 3.5% sodium chloride (alternate immersion).

Al-8Mg Aluminum Alloy: Effect of Additions of Various Elements on Stress-Corrosion Cracking, in Sodium Chloride

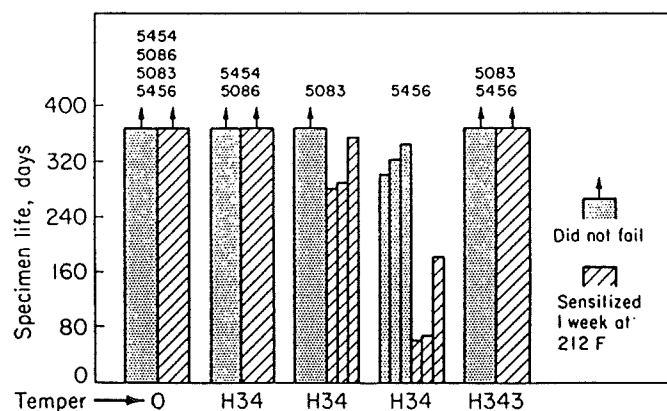


Effect of the addition of bismuth, chromium, and zirconium on stress-corrosion cracking of an Al-8Mg alloy.

The alloys were water quenched from 400 °C and sensitized at 130 °C for 7 days. They were tested at room temperature in 3.5% sodium chloride solution.

Al-Mg-Mn Alloys: Comparison of the Resistance to Stress-Corrosion Cracking, in Sodium Chloride

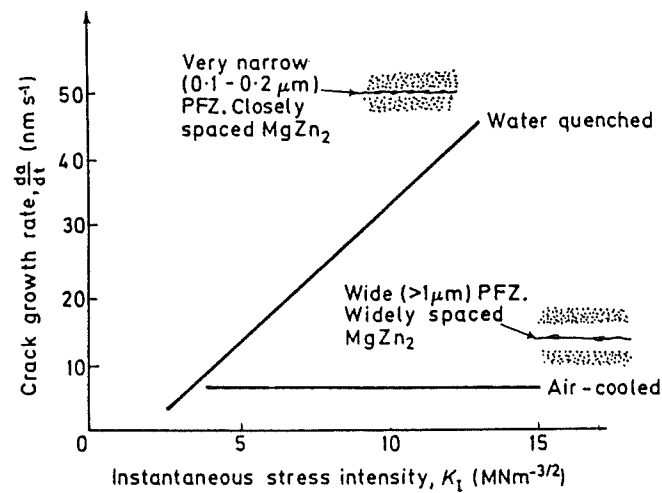
313



Comparison of the resistance to stress-corrosion cracking of several Al-Mg-Mn alloys exposed to 3.5% sodium chloride by alternate immersion.

In the O temper, alloys with up to 5.5% Mg (5456) have excellent resistance to stress-corrosion cracking. Strain-hardened and stabilized tempers, such as H34, have a low resistance when the magnesium content is in the range of 4.0 to 5.5%, as in alloys 5083 and 5456. Special tempers, such as H343, have been developed to provide a high resistance to stress-corrosion cracking.

Al-Zn-Mg Alloy: Variation of Stress-Corrosion Cracking Growth Rate with Stress-Intensity

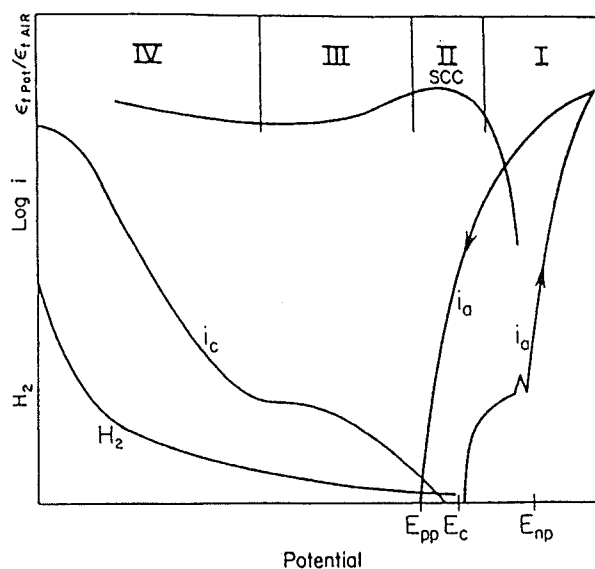


Variation of stress-corrosion cracking growth rate with stress-intensity of a high-purity Al-Zn-Mg alloy, water quenched or air cooled before aging.

The nature of the microstructure along the grain boundaries is indicated and depends on the cooling rate. The precipitate-free zone (PFZ) width increases as the cooling rate decreases, and the size and spacing of the MgZn₂ particles increases. The MgZn₂ particles are anodic with respect to the matrix.

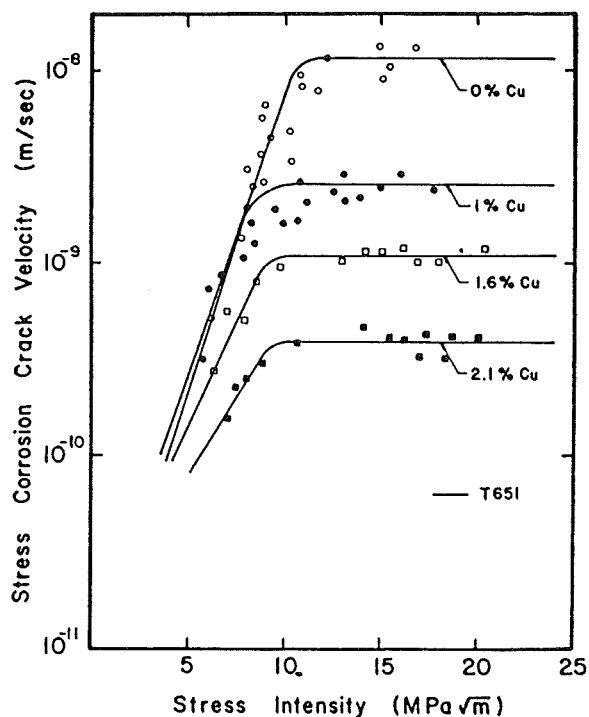
Al-Zn-Mg Alloys: Effects of Applied Potential on Stress-Corrosion Cracking Susceptibility, in Deaerated Solution

315



Effects of applied potential on stress-corrosion cracking susceptibility, anodic and cathodic current density, and hydrogen uptake by Al-Zn-Mg alloys in deaerated solution at room temperature.

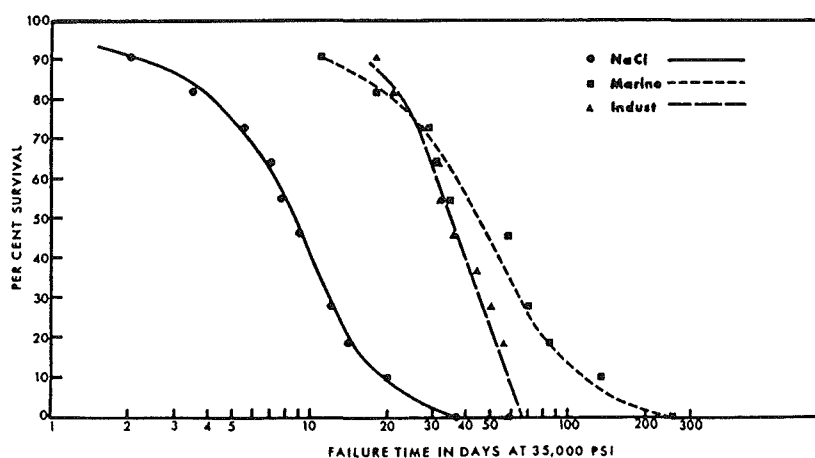
Al-Zn-Mg-XCu-T651 Alloys: Effect of Stress-Intensity on Stress-Corrosion Crack Velocity, in Sodium Chloride



Effect of stress-intensity on stress-corrosion crack velocity of Al-6Zn-2Mg-XCu alloys in 3.5% sodium chloride. The alloys were tested in the T651 condition. Note that the precipitate Co_2Al_9 is cathodic with respect to the aluminum matrix.

Al-4.25Zn-2.9Mg Aluminum Alloy: Stress-Corrosion Cracking Results in Various Environments

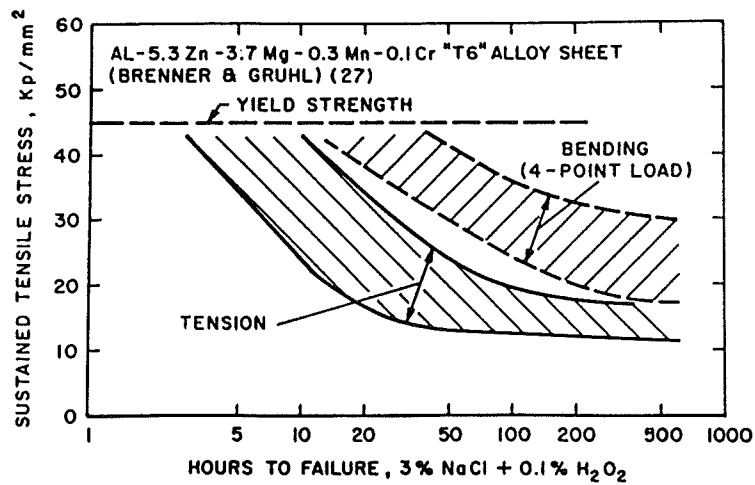
317



Stress-corrosion cracking results for alloy Al-4.25Zn-2.9Mg in a peak aged condition.

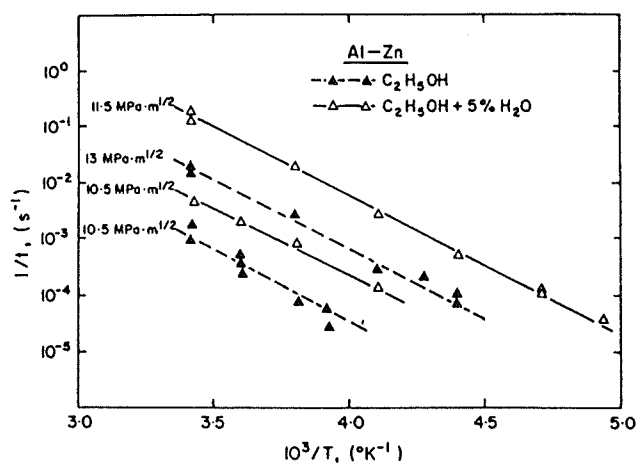
C-ring type specimens were used with the maximum tensile stress applied in the short-transverse direction. The percent survival time-to-failure at 35,000 psi in three environments is indicated. The specimens represent 11 lots of plate on which tests were started during the winter months.

Al-5.3Zn-3.7Mg-0.3Mn-0.1Cr-T6 Aluminum Alloy Sheet: Susceptibility to Stress-Corrosion Cracking



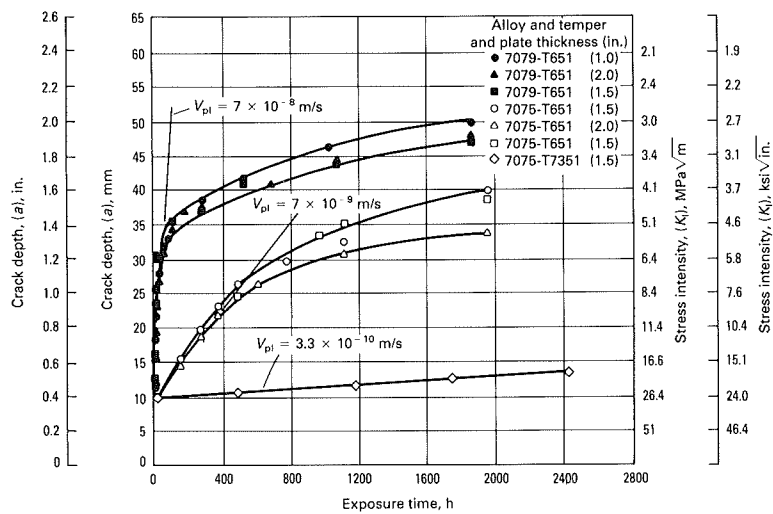
Comparison of direct tension loading with loading under bending on susceptibility to stress-corrosion cracking of an aluminum alloy.

Not only were lifetimes for specimens stressed in tension less than for those stressed in bending, but a lower threshold stress for stress-corrosion cracking was also indicated. This difference should be kept in mind when comparing results from tension tests and torsion tests. Furthermore, specimens tested under constant bending displacement would be expected to exhibit greater resistance to stress-corrosion cracking because of stress relaxation.



Arrhenius plot of crack initiation rates ($1/t$) versus reciprocal of temperature ($1/T$) for Al-21.5Zn alloy in reagent grade ethanol and in ethanol plus 5% water.

Various Aluminum Alloys: Crack Length and Stress-Intensity Versus Time Curves, in Sodium Chloride

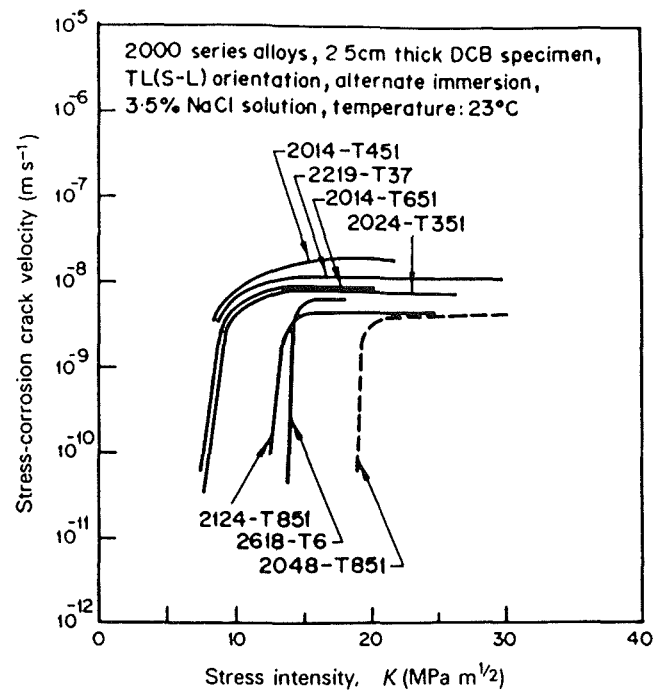


Crack length and stress-intensity versus time curves for double beam specimens of 7xxx series aluminum alloys, S-L orientation.

The specimens were double cantilever beam type, loaded to pop-in, and wetted three times daily with 3.5% sodium chloride. The figure indicates the nature of the data obtained in the plateau and the near-threshold regions in a decreasing-K test. In the event that an actual arrest does not occur, the threshold level is often taken to be the stress-intensity factor level at which the growth rate is about 10^{-10} m/s, i.e., where the growth rate is less than 0.2 mm in 30 days.

Various Aluminum Alloys: Crack Propagation Rates, in Aqueous Sodium Chloride

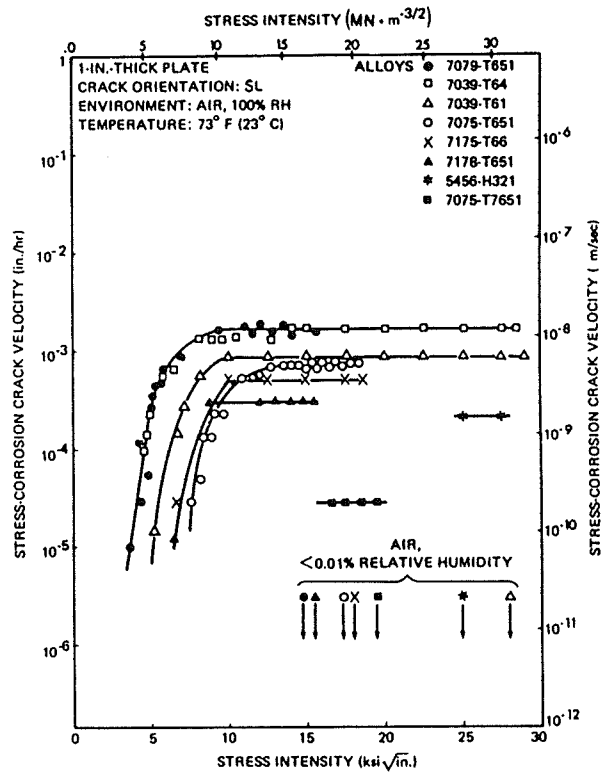
321



Crack propagation rates in stress-corrosion tests using pre-cracked specimens of 2xxx series high-strength aluminum alloys exposed to an aqueous solution of 3.5% sodium chloride.

The specimens were 25-mm thick, double cantilever beam type, loaded in the short-transverse direction. They were wet with solution twice a day.

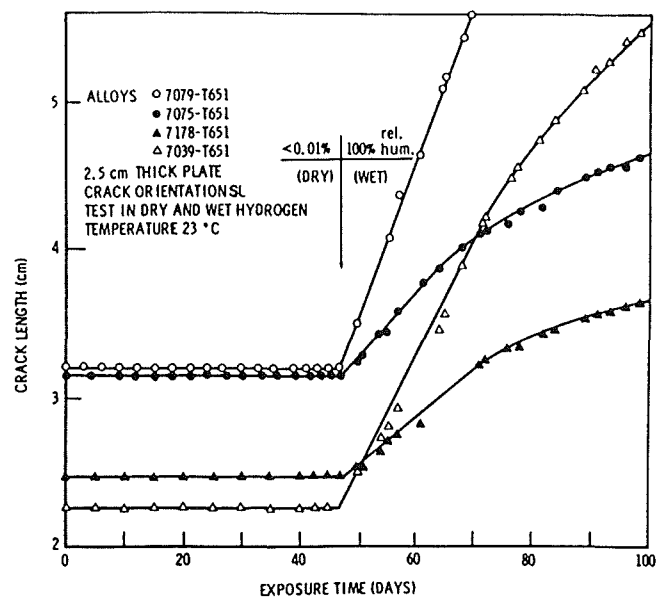
Various Aluminum Alloys: Effect of Moisture and Stress-Intensity on Subcritical Crack Growth, in Hydrogen Gas



Effect of moisture and stress-intensity on the subcritical crack growth rates of several high-strength aluminum alloys in air. The plateau crack velocities have the same order of magnitude as in moist hydrogen gas.

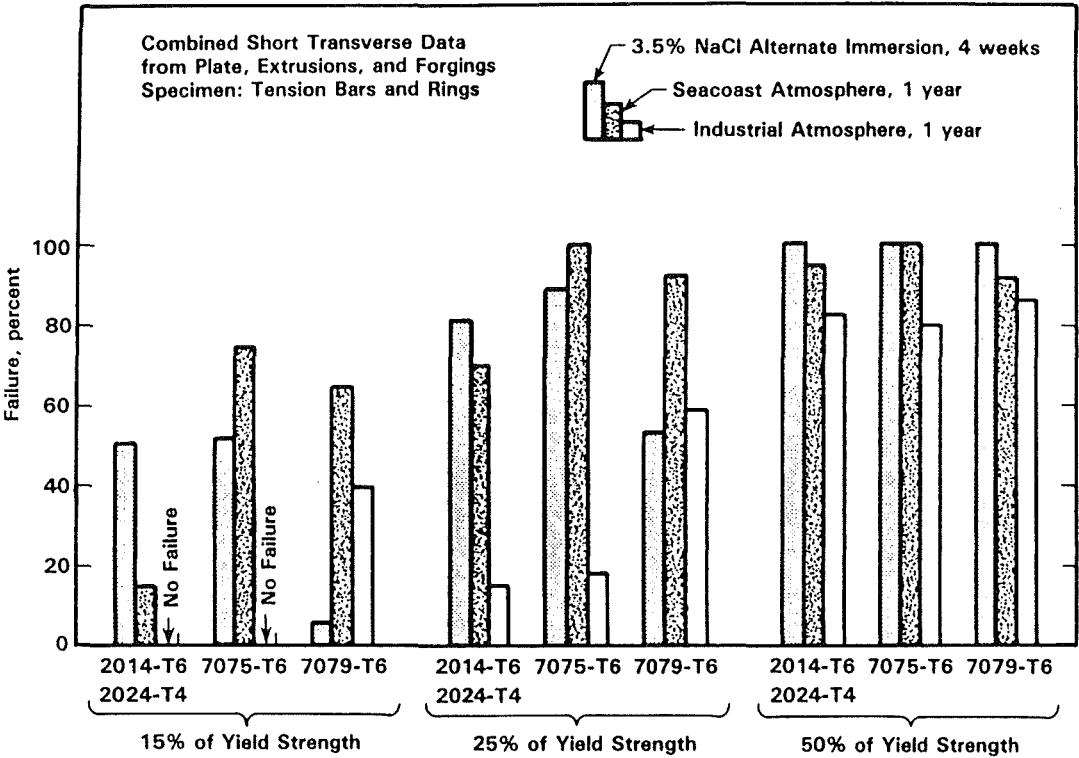
Various Aluminum Alloys: Effect of Humidity on Subcritical Crack Growth, in Hydrogen Gas

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Effect of humidity on subcritical crack growth of high-strength aluminum alloys in hydrogen gas. No crack growth was observed in dry hydrogen, whereas rapid crack growth was observed in wet hydrogen gas.

Various Aluminum Alloys: Probability of Failure by Stress-Corrosion Cracking in Several Environments

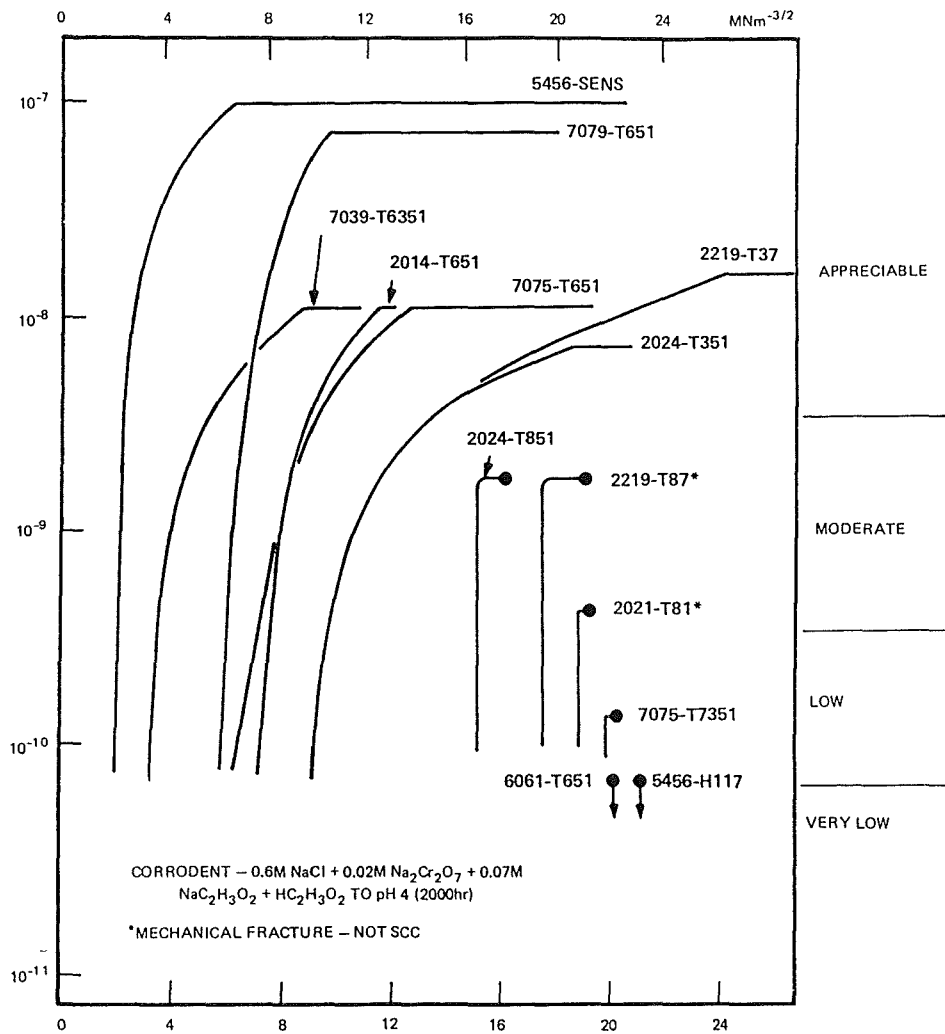


Comparison of the probability of failure by stress-corrosion cracking of smooth specimens in several environments.

Source: J.A. Beavers, G.H. Koch, and W.E. Berry, Corrosion of Metals in Marine Environments, Metals and Ceramics Information Center, Battelle Columbus Laboratories, July 1986.

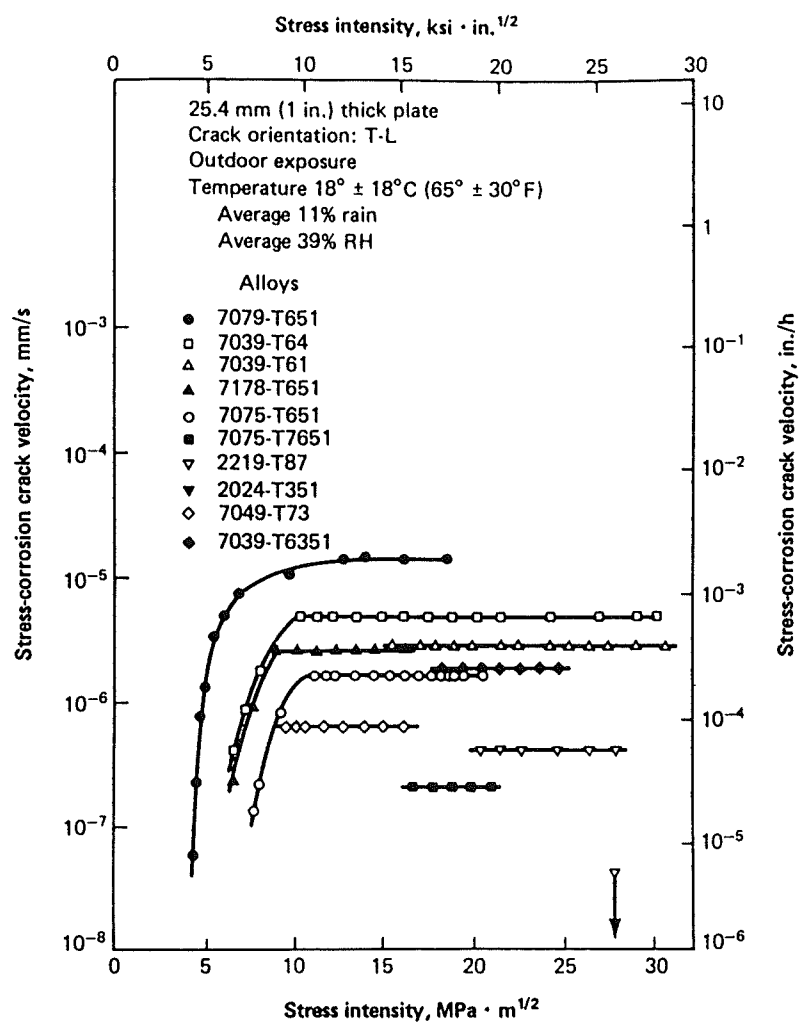
Various Aluminum Alloys: Stress-Corrosion Crack Growth as a Function of Stress-Intensity

325



Source: B.F. Brown, Stress Corrosion Cracking Control Measures, National Bureau of Standards Monograph 156, U.S. Department of Commerce, June 1977, p 28.

Various Aluminum Alloys: Effect of Outdoor Exposure and Stress-Intensity on Stress-Corrosion Crack Velocity

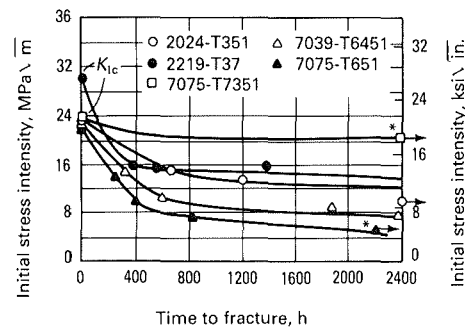


Effect of outdoor exposure and stress-intensity on the stress-corrosion crack velocity in several high-strength aluminum alloys.

Source: J.G. Kaufman and J.S. Santner, Fracture Properties of Aluminum Alloys, in Application of Fracture Mechanics for Selection of Metallic Structural Materials, J.E. Campbell, W.W. Gerberich, and J.H. Underwood, Ed., American Society for Metals, Metals Park, 1982, p 169-212.

Various Aluminum Alloys: Initial Stress-Intensity Versus Time-to-Fracture in Different Environments

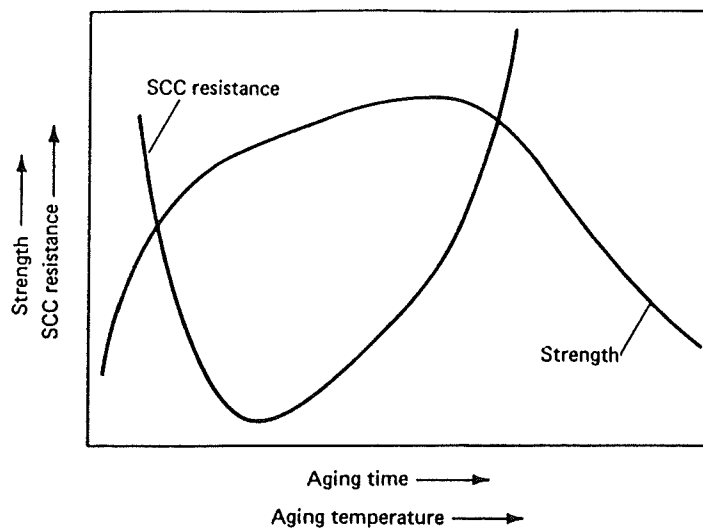
327



Initial stress-intensity versus time-to-fracture for compact specimens in the S-L plate orientation for various aluminum alloys.

The specimens were exposed to an aqueous solution 0.06M sodium chloride, 0.2M sodium dichromate, 0.07M sodium acetate, and acetic acid to pH 4. An asterisk indicates that stress-corrosion cracking had started.

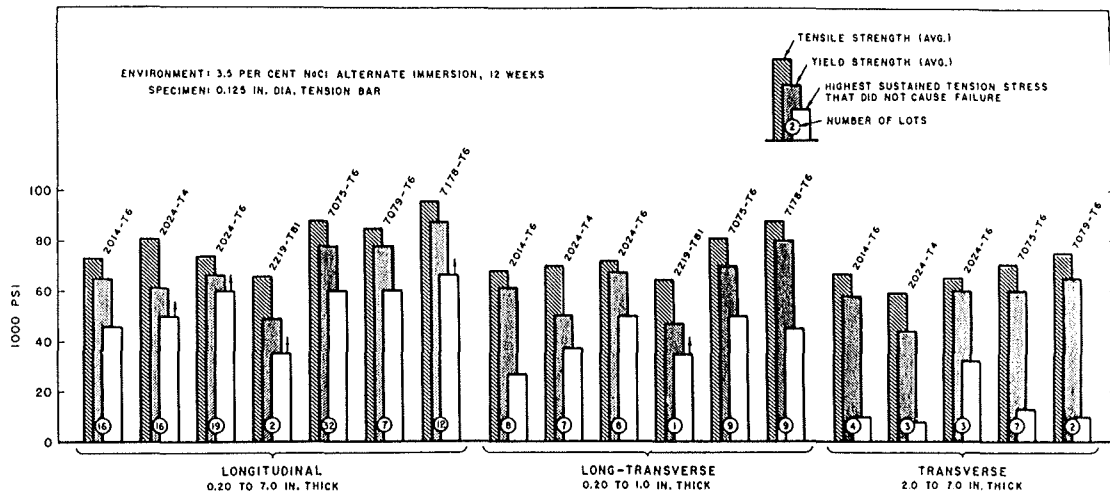
Various Aluminum Alloys: Variations in Strength and Stress-Corrosion Resistance as a Function of Aging Time and Temperature



Variations in strength and stress-corrosion resistance as a function of aging time and temperature for precipitation-hardening aluminum alloys.

Various Aluminum Alloys: Resistance of Extrusions to Stress-Corrosion Cracking

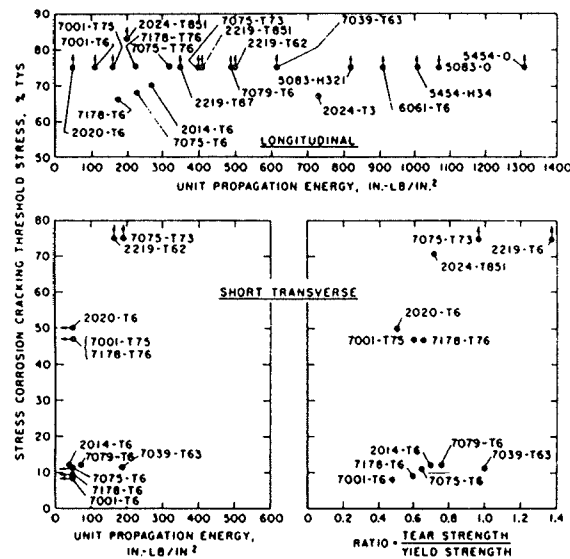
329



Comparison of the relative resistance of extrusions in several high-strength aluminum alloys to stress-corrosion cracking.

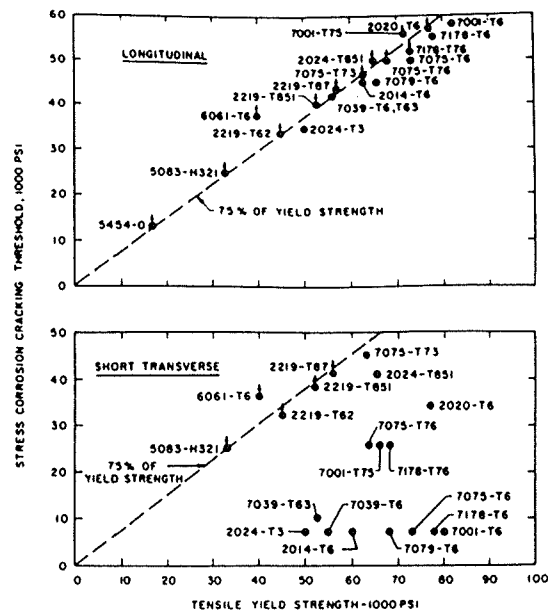
The arrows indicate that no stress-corrosion cracking failures occurred at the highest stress employed. The tensile and yield strengths for each alloy are also shown as a function of orientation in the extrusion. The lowest resistance to stress-corrosion cracking is in the short-transverse direction.

Aluminum Alloy Plate: Correlation of Resistance with Fracture Toughness, in Sodium Chloride



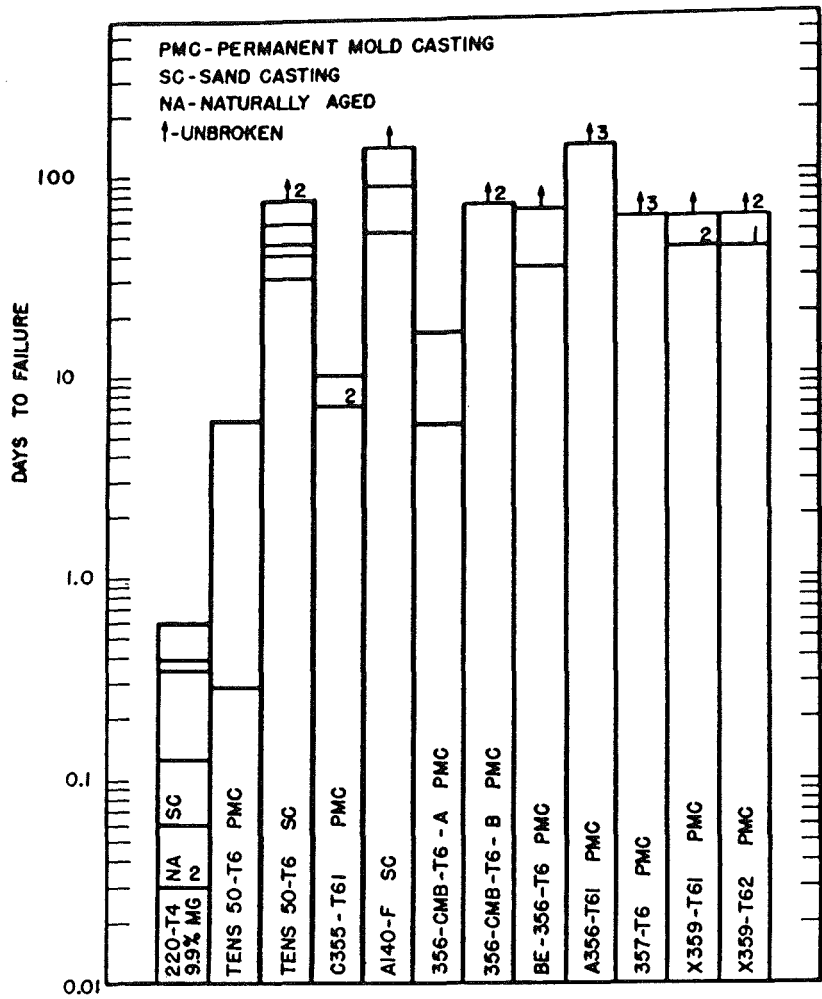
Correlation of resistance to stress-corrosion cracking with fracture toughness of aluminum alloy plate. The stress-corrosion threshold data were obtained with exposures of 84 days or longer to the 3.5% sodium chloride alternate immersion test.

Source: J.G. Kaufman and J.S. Santner, *Fracture Properties of Aluminum Alloys*, in *Application of Fracture Mechanics for Selection of Metallic Structural Materials*, J.E. Campbell, W.W. Gerberich, and J.H. Underwood, Ed., American Society for Metals, Metals Park, 1982, p 169-212.



Threshold stress for stress-corrosion cracking as a function of yield strength for a wide variety of aluminum alloys and tempers. The stress-corrosion threshold data were obtained with exposures of 84 days or longer to the 3.5% sodium chloride alternate immersion test.

Cast Aluminum Alloys: Times-to-Failure by Stress-Corrosion Cracking

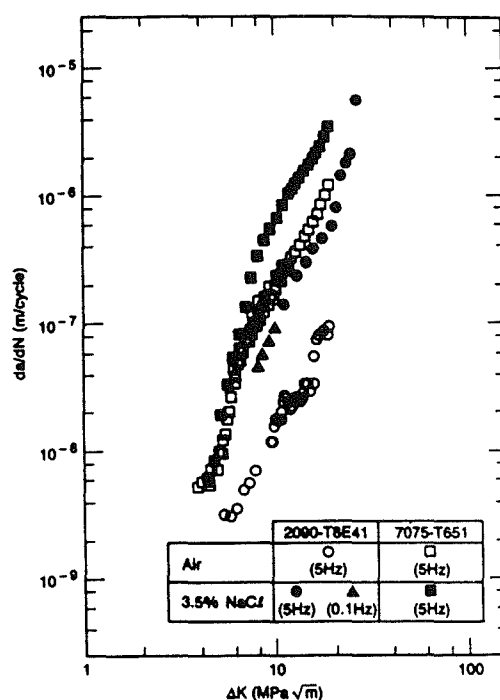


Times-to-failure by stress-corrosion cracking of cast aluminum alloys. Horizontal lines in a column signify failure of a specimen. Numbers on these horizontal lines indicate the number of specimens that failed. Arrows signify specimens removed unbroken. Stress level 75% of yield strength. Environment: Specimens immersed for 3 min in each 45 in a 1N NaCl solution containing 10 ml/L of 30% hydrogen peroxide.

Source: H.L. Logan, The Stress Corrosion of Metals, John Wiley & Sons, New York, 1966, p 209.

2090-T8E41 and 7075-T651 Aluminum Alloys: Comparison of the Rate of Fatigue Crack Growth as a Function of the Range of the Stress-Intensity Factor, in Sodium Chloride

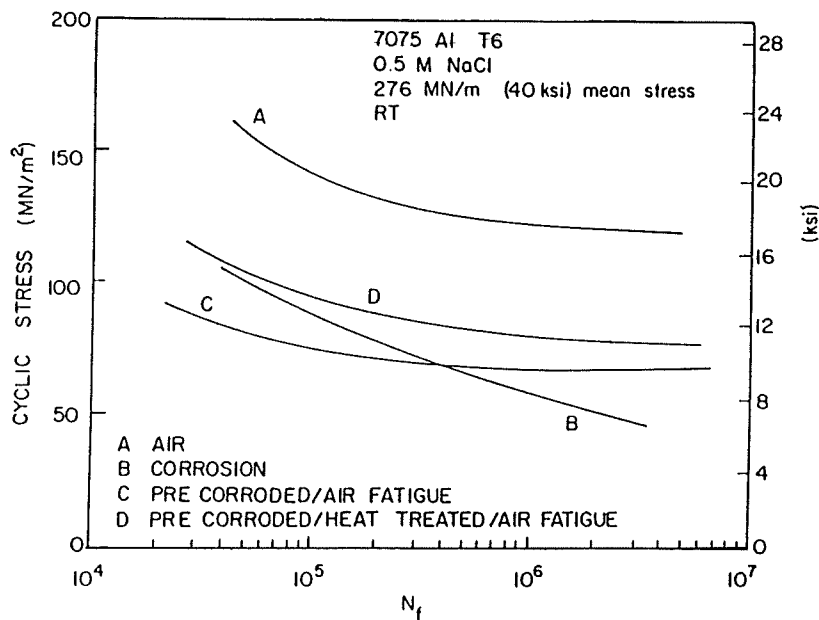
333



Comparison of the rate of fatigue crack growth as a function of the range of the stress-intensity factor for an Al-Li alloy, 2090-T8E41, and 7075-T651 in air and in 3.5% sodium chloride solution.

The specimens were tested in the L-T orientation. In air, the Al-Li alloy exhibited a very tortuous fracture surface morphology and high crack closure levels that led to fatigue crack growth rates an order of magnitude lower than for the 7075 alloy. Such roughness-induced crack closure was greatly reduced for specimens tested in 3.5% sodium chloride solution, and the fatigue crack growth rates were elevated six-fold over those in air. Despite this increase in rate, the corrosion fatigue crack growth rate for the Al-Li alloy was still less than for the 7075 alloy.

7075-T6 Aluminum Alloy: Effect of Corrosion and Precorrosion on Fatigue Lives

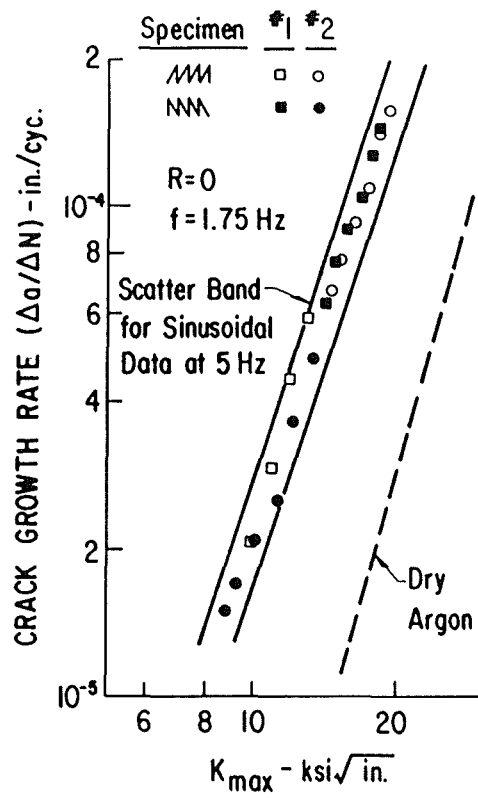


Effects of corrosion and precorrosion on the fatigue lives of 7075-T6 alloy.

Note that re-solutionizing and re-aging the alloy after pre-exposure has resulted in a significant increase in fatigue resistance, perhaps because of the elimination of hydrogen that had been picked up during precorrosion.

7075-T651 Aluminum Alloy: Effect of Cyclic Waveform on Fatigue Crack Growth, in Distilled Water

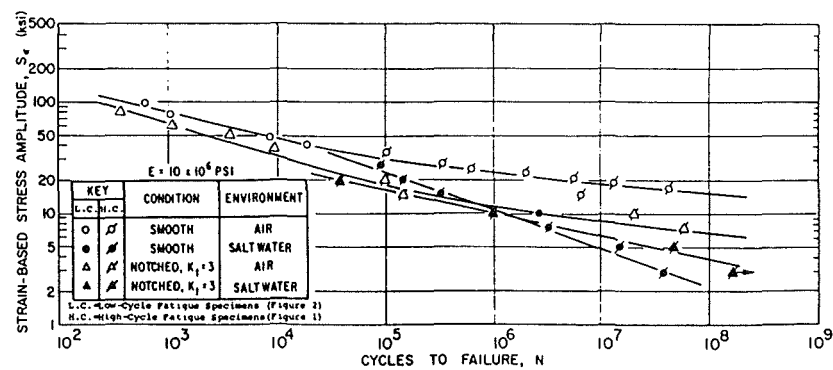
335



Effect of cyclic waveform on fatigue crack growth in a 7075-T651 aluminum alloy tested in distilled water at room temperature.

In contrast to results obtained by Barsom on steels and by Pelloux on 7075-T6 and 7075-T73, a positive sawtooth waveform did not result in a more rapid rate of corrosion fatigue crack growth as compared to a negative sawtooth waveform. In all cases, the maximum K level was below K_{Isc} . The reason for this discrepancy is not understood.

7079-T6 Aluminum Alloy: Flexural Fatigue Curves for Smooth and Notched Specimens

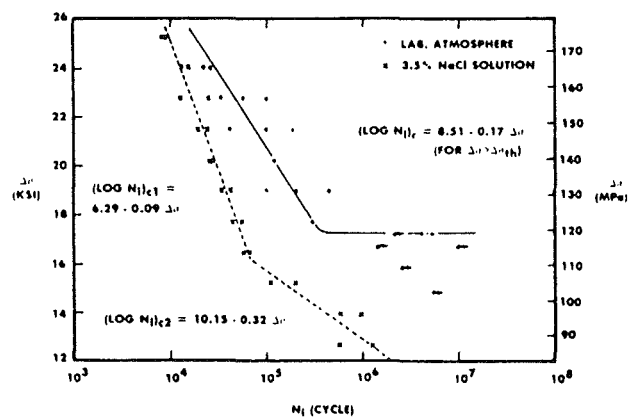


Flexural fatigue curves for smooth and notched specimens of aluminum alloy 7079-T6 as a function of environment.

Source: R.W. Hertzberg, Deformation and Fracture Mechanics of Engineering Materials, 2nd ed., John Wiley & Sons, New York, 1983, p 446-447.

7475-T7351 Aluminum Alloy: Influence of Environment on Number of Cycles to Crack Initiation

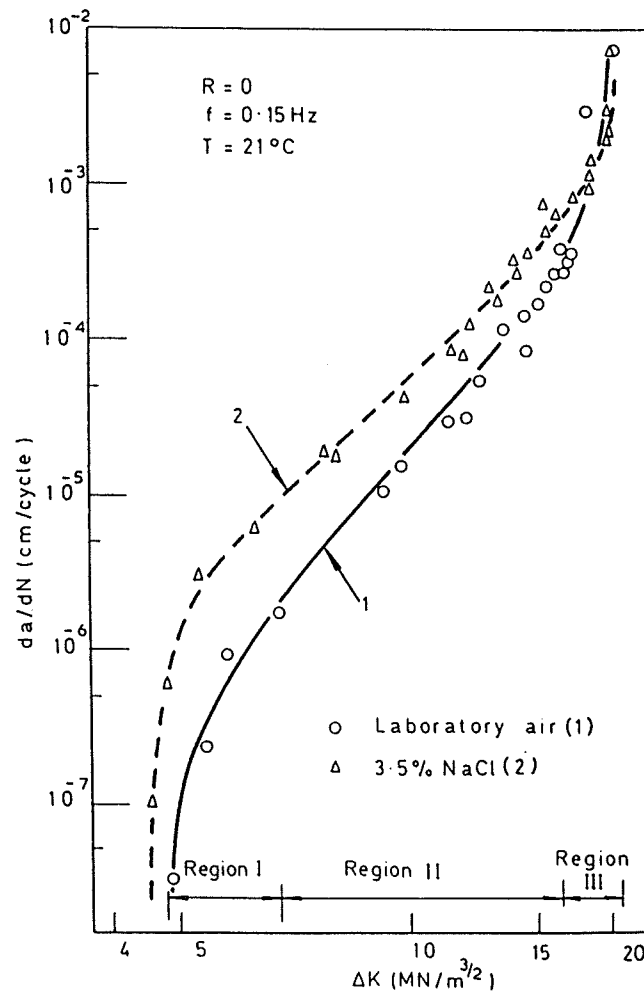
337



Influence of environment on the number of cycles to crack initiation as a function of stress level for aluminum alloy 7475-T7351.

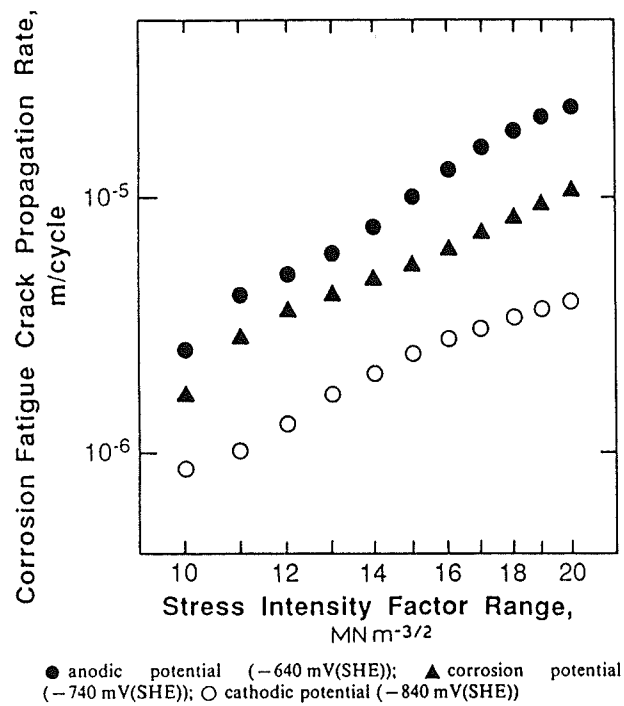
Each specimen contained a hole. The fatigue crack initiation site was located at the edge of this hole.

RR58 Aluminum Alloy: Fatigue Crack Growth Rate, in Air and in Sodium Chloride



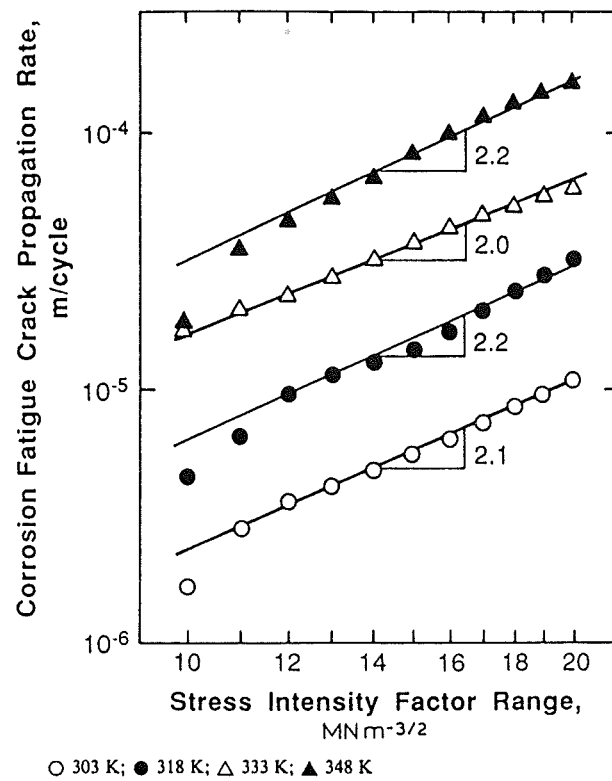
Comparison of the rate of fatigue crack growth in air and in 3.5% sodium chloride for aluminum alloy RR58.

Composition: 1.8 to 2.7% Cu, 1.2 to 1.8% Mg, 0.15 to 0.25% Si, 0.9 to 1.4% Fe, 0.8 to 1.4% Ni, bal Al.

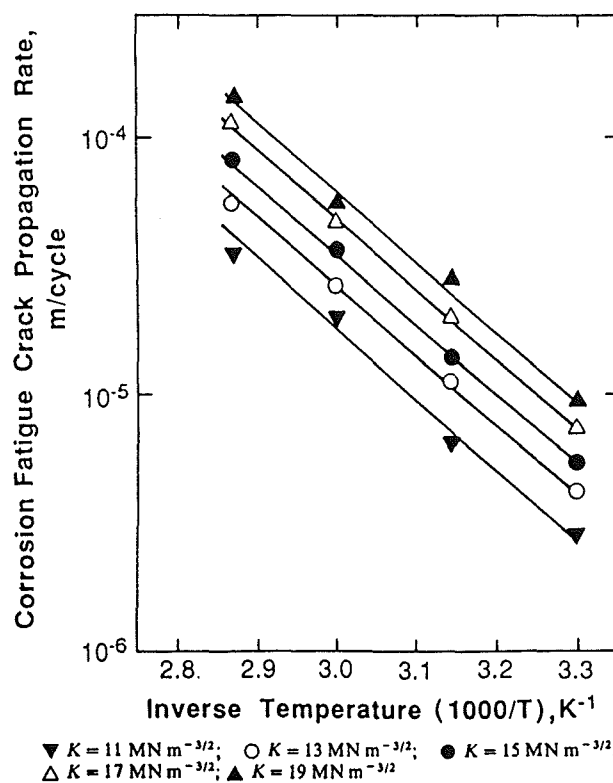


Corrosion fatigue crack propagation rate of a high-purity Al-Zn-Mg alloy versus stress-intensity range for various applied potentials.

Al-Zn-Mg Alloy: Corrosion Fatigue Crack Propagation Rate, in 3.5% NaCl Solution



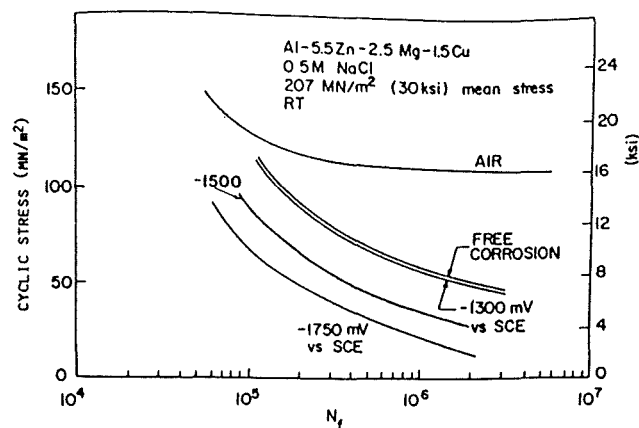
Corrosion fatigue crack propagation rate for high-purity Al-Zn-Mg alloy versus stress-intensity range at corrosion potential ($-840 \text{ mV}_{\text{SHE}}$) for various temperatures. ○ 303 K; ● 318 K; △ 333 K; ▲ 348 K.



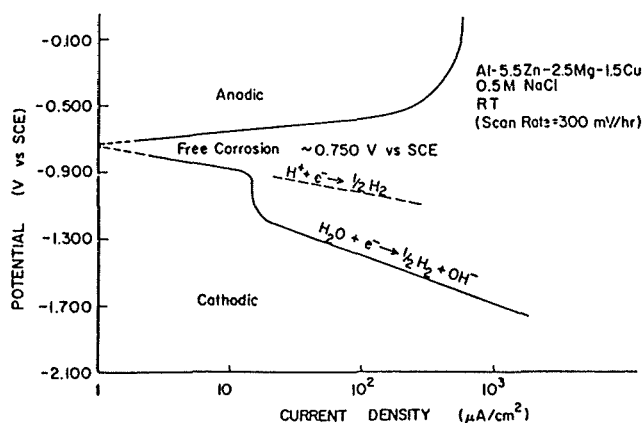
Arrhenius-type plots of corrosion fatigue crack propagation rate for high-purity Al-Zn-Mg alloy at a corrosion potential of -740 mVSHE.

Activation energy $\sim 52 \text{ kJ/mol}$. Results deviated from simple superposition model.

Al-5.5Zn-2.5Mg-1.5Cu Aluminum Alloy: S-N Curves, in Sodium Chloride Solutions



(a)



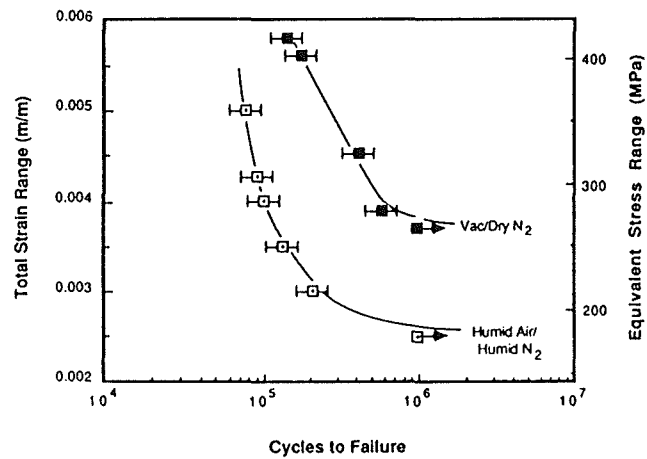
(b)

S-N curves. (a) Comparison of the S-N behavior of polycrystalline Al-5.5Zn-2.5Mg-1.5Cu alloy heat treated to the T-6 condition in dry air with S-N results obtained as a function of applied cathodic potentials in aerated 0.5M sodium chloride. (b) Polarization curve of polycrystalline Al-5.5Zn-2.5Mg-1.5Cu alloy in 0.5M sodium chloride solutions.

The S-N curves in the aerated sodium chloride solution are essentially the same under free corrosion conditions and at -1300 mV_{SCE}. It is possible that oxygen reduction rather than hydrogen reduction is involved in this range. Hydrogen reduction may become dominant at greater cathodic potentials. There was also a change from intergranular to transgranular fracture as the cathodic potential was increased. It was suggested that both stress-corrosion cracking and corrosion fatigue in this alloy are due to hydrogen embrittlement.

Al-5.6Zn-1.94Mg Aluminum Alloy: Cycles-to-Failure, in Nitrogen Gas

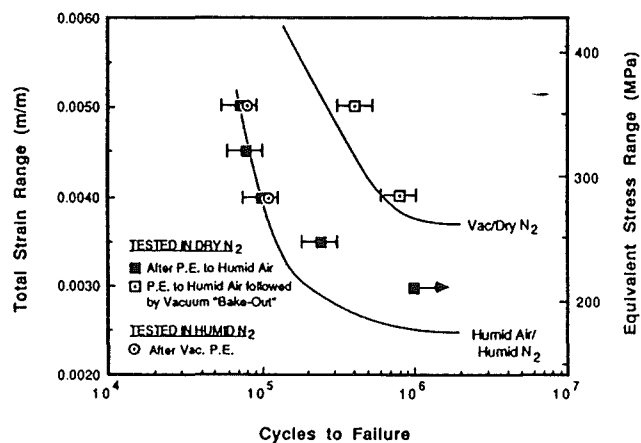
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Cycles-to-failure for vacuum outgassed samples tested in dry nitrogen gas and for samples pre-exposed to humid air tested in humid nitrogen gas.

The material tested was a high-purity ternary alloy, Al-5.6Zn-1.94Mg, which had been aged to peak hardness.

Al-5.6Zn-1.94Mg Aluminum Alloy: Effect of Pre-exposure and Test Environment on Fatigue Life

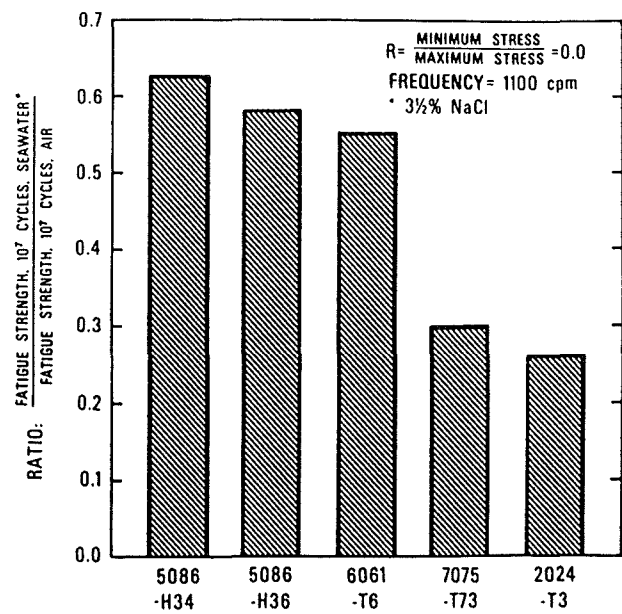


Effect of different pre-exposure and test environment combinations on fatigue life as compared to the results shown in the previous figure (solid lines).

These results show that the reduction in lifetime of Al-Zn-Mg alloys in water vapor was due to hydrogen embrittlement.

Aluminum Alloy Sheet: Axial Stress Fatigue Strengths, in Air and Seawater

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Axial stress fatigue strengths of aluminum alloy sheet in air and in seawater.

COPPER ALLOYS

COPPER ALLOYS

Stress-corrosion cracking (SCC) of copper alloys has been found to occur in a wide variety of environments. A generic classification of these alloys is given below. In some cases, the conditions for cracking are very limited and exist only within a narrow range of pH values, or a narrow range of potentials. Specific environments in which cracking has been observed include:

Acetate solution. Pure copper wire stressed beyond yield and the copper-zinc alloy C26000, in 0.05 N cupric acetate.

Amines. Alloy C26000 is susceptible to SCC in solutions of methyl amine, ethyl amine, and butyl amine when dissolved copper is present in the solution. Tubing fabricated from C68700 exhibited SCC from the steam side of a condenser system in a desalinization plant, probably because an amine was used as a water treatment chemical.

Ammonia. All copper-base alloys can be made to crack in NH_3 vapor, NH_3 solutions, ammonium ion solutions, and environments in which NH_3 is a reaction product. The rate at which cracks develop is dependent on many variables including the stress level, the specific alloy, pH, NH_3 or NH_4 concentration, copper ion concentration, and potential.

Atmosphere. Many environments contain pollutants which may cause SCC problems in the presence of moisture. Sulfur dioxide, oxides of nitrogen, chlorides, and NH_3 are known to induce SCC in some copper alloys.

Chlorate solutions. Brass was observed to crack intergranularly and transgranularly when immersed in 0.1 to 0.5 M sodium chlorate (NaClO_3) solutions at pHs from 3.5 to 9.5.

Chloride solutions. Slow strain rate tests have shown that the alloys C26000 and C44300 have lower fracture stresses in NaCl solutions when the metal is anodically polarized. In the absence of an applied potential there is no significant change in fracture stress relative to in-air.

Citrate solutions. Alloy C72000 is sensitive to intergranular SCC in citrate solutions containing dissolved copper in the pH range 7 to 11.

Formate solutions. Brass is susceptible to SCC in sodium formate (NaCHO_2) solutions at pHs exceeding 11 over a considerable range of applied potentials.

Hydroxide solutions. Brass exhibits SCC under slow strain rate testing when exposed to NaOH at pHs of 12 and 13.

Mercury and mercury salt solutions. Copper alloys crack readily when under stress—either applied or residual—they are exposed to metallic mercury or mercury salt solutions that deposit mercury on the surface of the alloy. This high sensitivity to mercury has a positive aspect in that it is the basis for an industrial test for the detection of residual stresses in which the component is immersed in a solution of mercurous nitrate. Liquid metal embrittlement may be considered to be a special form of SCC.

Nitrate solutions. Transgranular cracking of alloy C44300 has been observed when immersed in naturally aerated 1 N sodium nitrate (NaNO_3) solution at pH 8 and a potential of 0.15 V vs standard hydrogen electrode (SHE).

Nitrite solutions. 99.9 and 99.996% pure copper exhibited transgranular cracking when subjected to a strain rate of 10^{-6} s^{-1} while immersed in 1 M sodium nitrite (NaNO_2) at a pH of 8.2. Cracking in M NaNO_2 was also observed in C26000, admiralty brasses, and C70600.

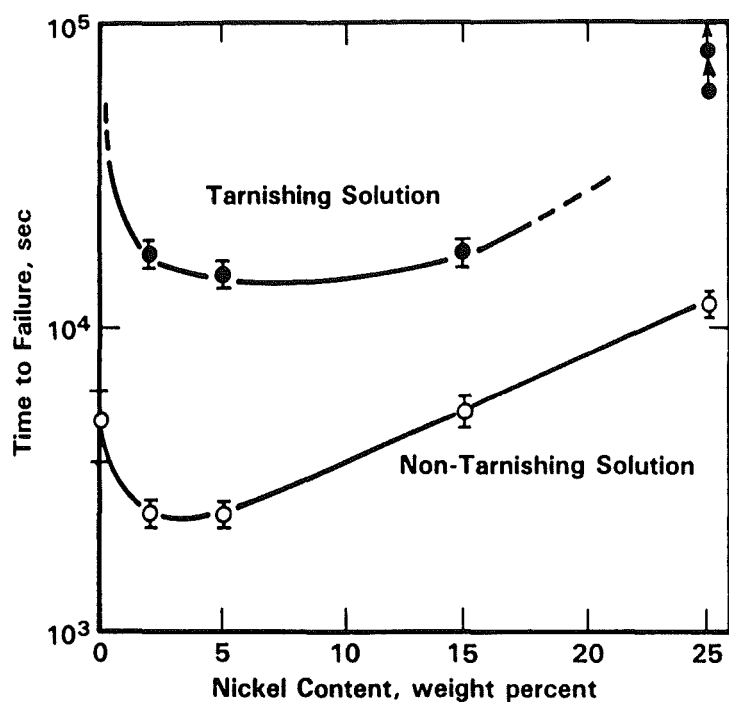
Solder. The application of lead-containing solders at 400° F to copper alloys can result in cracking if the alloy is deformed immediately following the application of the solder.

Sulfur dioxide. Brass is susceptible to SCC in moist air containing 0.05 to 0.5 vol% SO_2 .

Sulfate solutions. SCC of C26000 was observed in a solution of 1 N sodium sulfate (Na_2SO_4) and 0.01 N H_2SO_4 when the alloy was polarized at a potential of 0.25 V vs SHE and subjected to a constant strain.

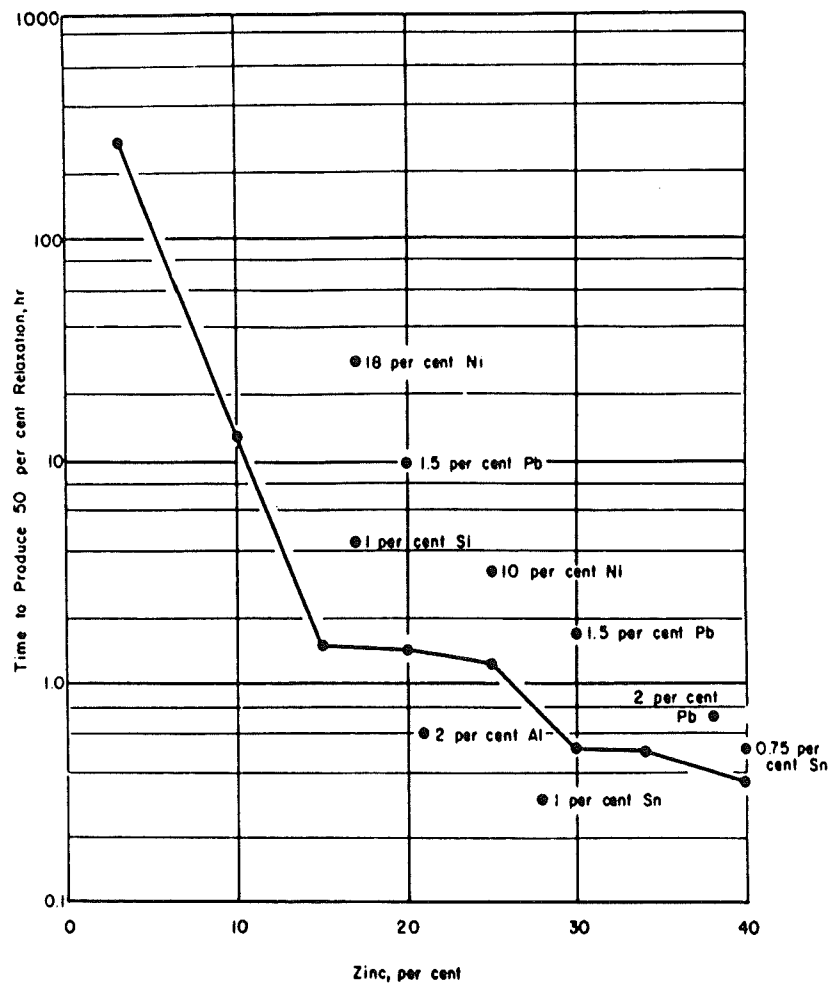
Tungstate solutions. Mild transgranular SCC of C44300 was observed in 1 N sodium tungstate (Na_2WO_4) at pH 9.4 and a corrosion potential of 0.080 V vs SHE.

Water. Several cases of SCC of admiralty brass heat-exchanger tubing has occurred in stagnant water, stagnant water contaminated with NH_3 , and water accidentally contaminated with a nitrate.



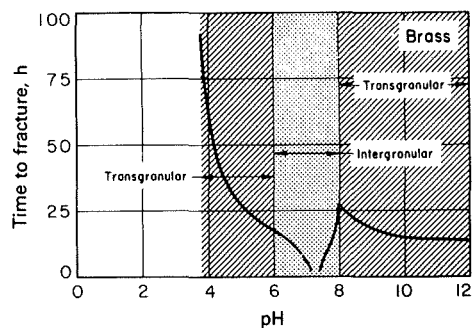
Effect of nickel content on the time-to-failure of copper-nickel alloys tested in tarnishing and nontarnishing 15*N* aqueous ammonia.

Cu-Zn Alloys: Stress-Corrosion Cracking



Stress-corrosion cracking of Cu-Zn alloys in a moist ammoniacal atmosphere.

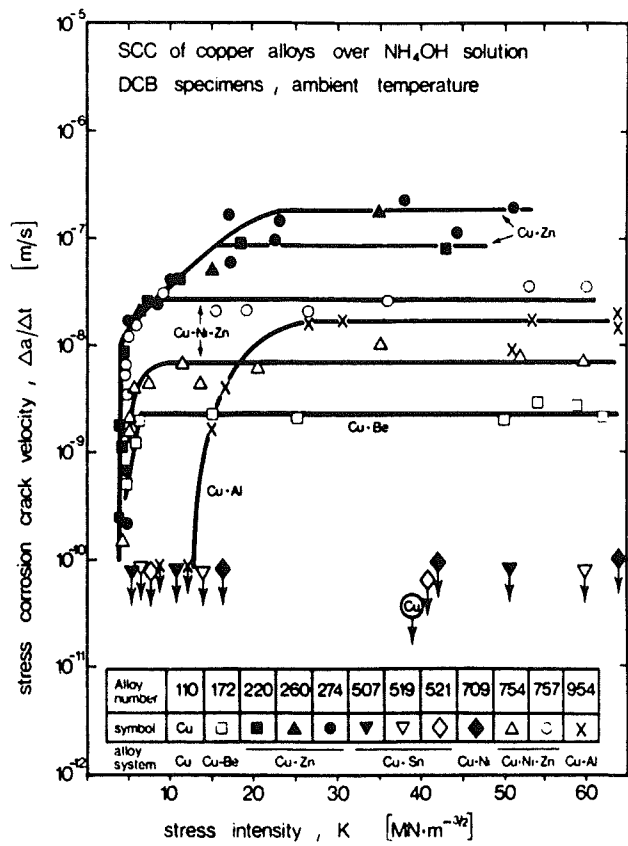
Source: Metals Handbook, Corrosion, Vol 13, 9th ed., ASM International, Metals Park, 1987.



Effect of pH on the time-to-fracture of brass by stress-corrosion cracking in ammoniacal copper sulfate solution at room temperature.

In aqueous solutions, the pH level has a strong influence on whether crack paths are intergranular or transgranular. Cracking occurs most rapidly in nearly neutral solutions, where the crack path is intergranular. The crack path is transgranular in both alkaline and acidic solutions, and brass is highly resistant to cracking when the pH is less than 4.

Cu-Zn and Cu-Ni Alloys: Rate of Stress-Corrosion Cracking, in Ammonia Atmosphere



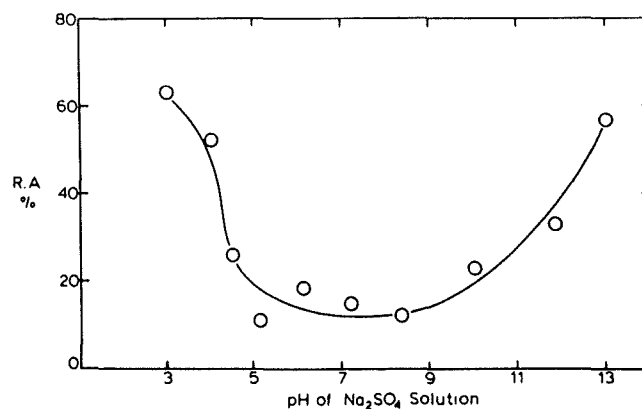
Rate of stress-corrosion cracking of commercial copper-base alloys in ammonia atmosphere.

Note that copper-zinc alloys crack the fastest, whereas copper-nickel alloys are entirely stress-corrosion cracking resistant, in accordance with service experience in condensers.

Source: M.O. Speidel, Stress Corrosion Cracking and Corrosion Fatigue Fracture Mechanics, in Corrosion in Power Generating Equipment, M.O. Speidel and A. Atrens, Ed., Plenum Press, New York, 1984, p 85-132.

Cu-42Zn Alpha-Beta Brass Alloy: Ductility as Measured by Reduction in Area Versus Solution pH, in Na₂SO₄

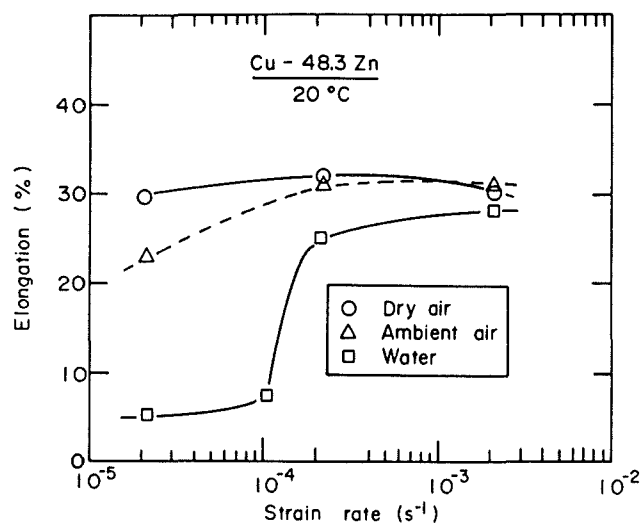
355



Ductility of a Cu-42Zn α - β brass alloy as measured by reduction in area versus solution pH in 1N Na₂SO₄.

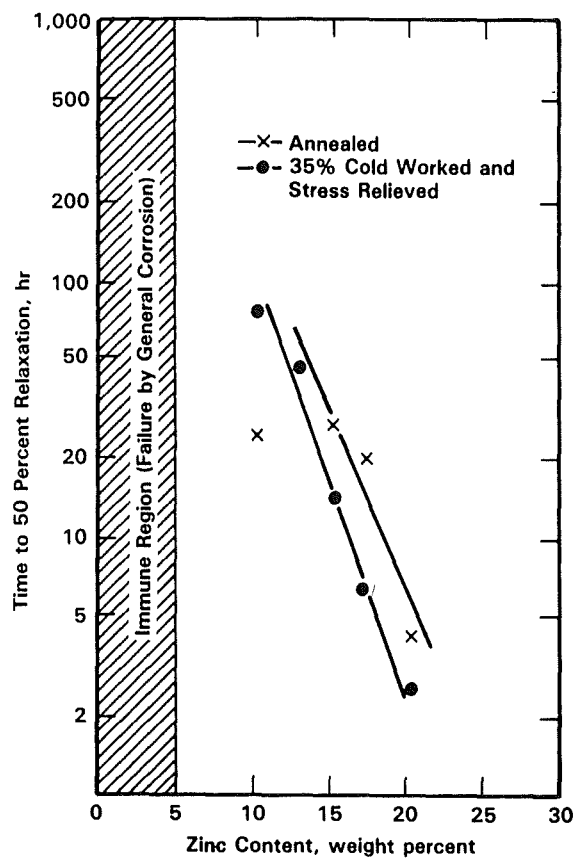
The extension rate was 5×10^{-6} cm/s, which corresponded to an initial strain rate of 4×10^{-6} /s.

Cu-48.3Zn Copper Alloy: Effect of Environment and Strain Rate on Elongation-to-Fracture



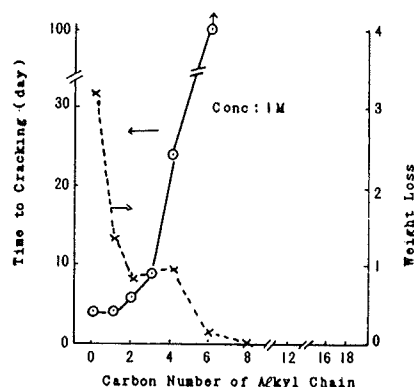
Effect of environment and strain rate on elongation-to-fracture of a β' brass, Cu-48.3Zn.

β' brass is an ordered bcc structure of the CsCl(2) type. It was observed that decreasing the electron/atom (e/a) ratio of the β' phase promoted transgranular cracking and increased the crack growth rates and that the most rapid crack growth rates were associated with alloy compositions that gave rise to a strain-induced martensitic transformation during stress-corrosion cracking.



Effect of zinc content on the resistance of Cu-10Ni-Zn alloys to stress-corrosion in an ammoniacal environment.

Yellow Brass (65Cu-35Zn): Time-to-Cracking and Weight Loss, as a Function of the Carbon Number of the Alkyl Chain, in Alkylamine Solution

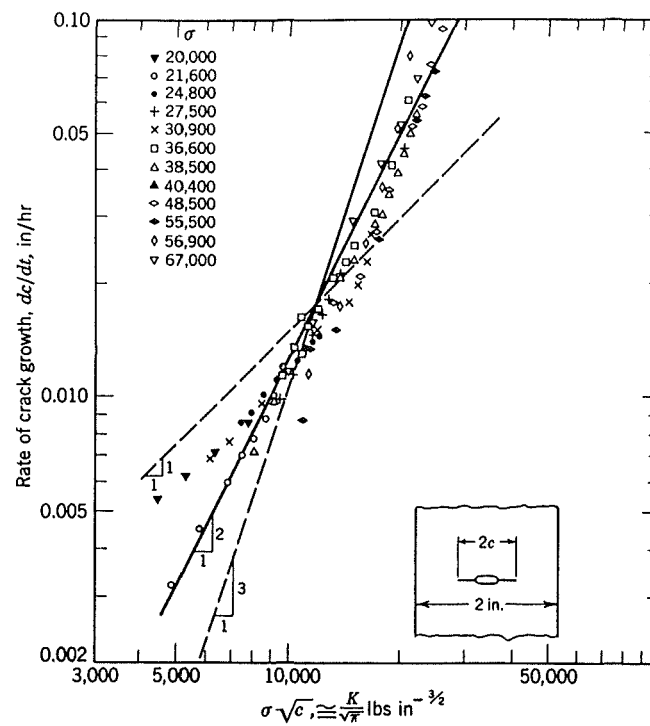


Time-to-cracking and weight loss of yellow brass (65Cu-35Zn) in 1M alkylamine solution as a function of the carbon number of the alkyl chain. The specimens were deep drawn cups in the as-drawn condition.

The paper describes the influence of approximately 200 chemicals on stress-corrosion cracking of yellow brass.

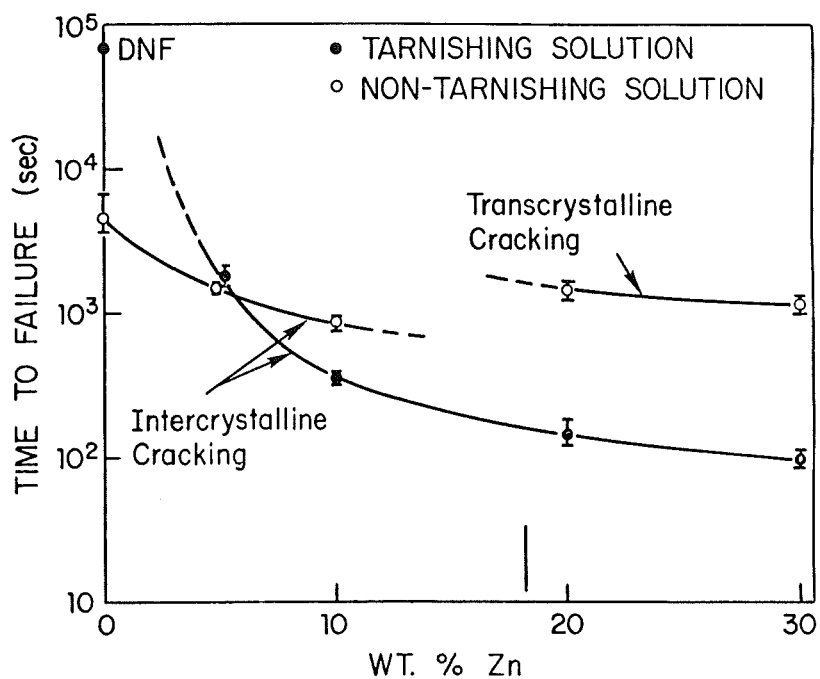
**Brass: Stress-Corrosion Crack Propagation Rate, in
CuSO₄ + (NH₄)₂SO₄**

359

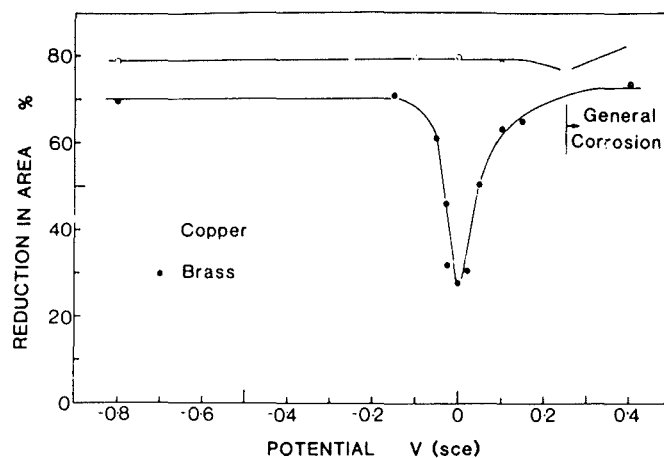


**Rate of stress-corrosion crack propagation in cold rolled brass
exposed to 0.05M CuSO₄ + 0.48M (NH₄)₂SO₄ (pH 7.25).**

Brass: Effect of Zinc Content on Time-to-Failure, in Ammoniacal Solutions



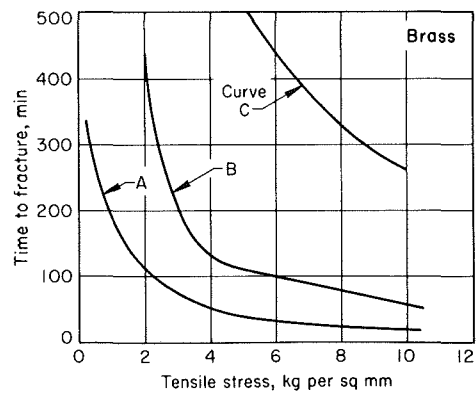
Effect of zinc content on the time-to-failure of brass specimens tested at 10% proof stress in tarnishing and nontarnishing 15N ammoniacal solutions.



Reduction in area as a function of potential showing maximum susceptibility to stress-corrosion cracking of 0.002% As brass (70-30) around 0 mV_{SCE} in double deionized water, but no stress-corrosion cracking in 99.99% Cu throughout the potential range from -800 to +400 mV_{SCE}.

In this study, which involved other environments as well, the results were interpreted as indicative of dezincification being involved in the various environments that promoted cracking, but the extent of dezincification plays an important role in determining whether or not cracking occurs, those additions leading to rapid and massive dezincification involving redeposition of dissolved copper not being conducive to cracking.

Brass: Effect of Tensile Stress on Time-to-Fracture, in Three Corrosive Environments

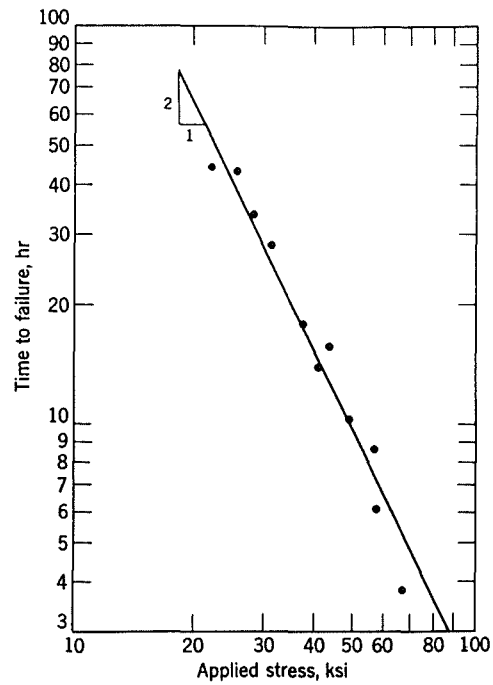


Effect of initial tensile stress on the time-to-fracture by stress-corrosion cracking of brass in three corrosive environments.

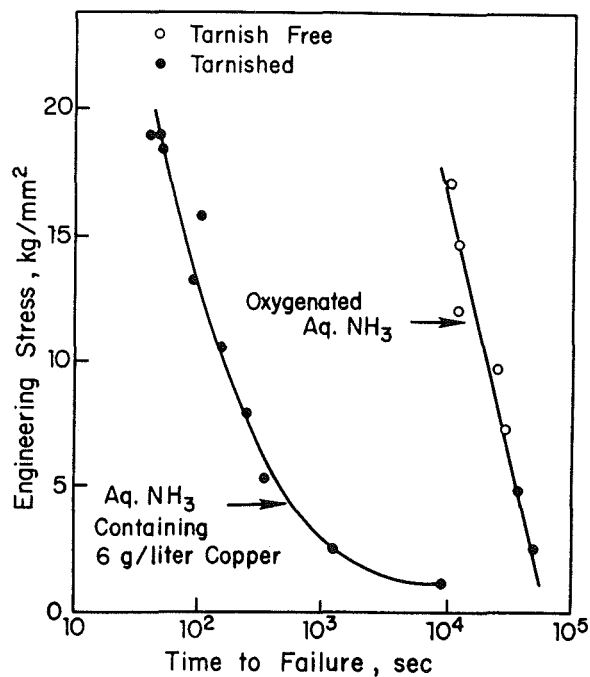
Curve A: Partly immersed in concentrated ammonium hydroxide. Curve B: Exposed to the vapor of concentrated ammonium hydroxide. Curve C: Exposed to a gaseous mixture of ammonia, oxygen, carbon dioxide, and water vapor.

70-30 Brass: Time-to-Rupture as a Function of Applied Stress, in Ammoniacal Solutions

363

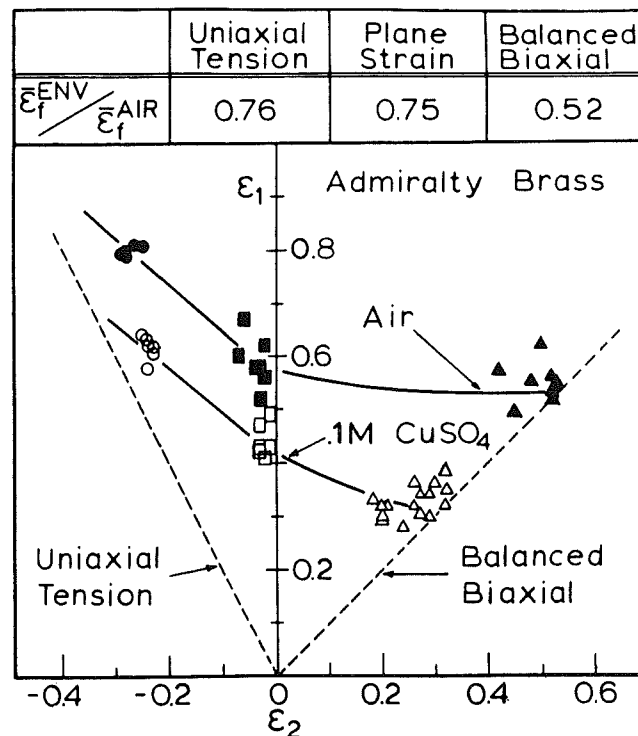


Time-to-rupture as a function of applied stress for notched cold rolled 70-30 brass specimens exposed to ammoniacal solution.



Stress-corrosion data for 70-30 brass specimens stressed in fresh or pre-concentrated 15*N* aqueous NH_3 .

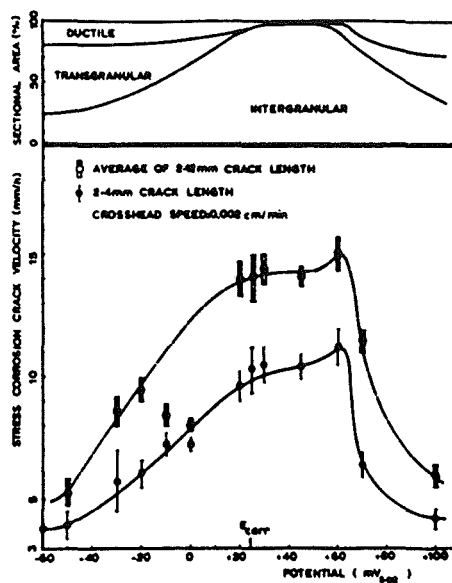
The copper exists in the pre-concentrated solutions as $\text{Cu}(\text{NH}_3)_5^{2+}$ ions. The presence of these complex ions imparts a blue color to the solution. If these ions play a role in the stress-corrosion process, the differences in time-to-failure shown may be attributed to the fact that time is required in the fresh solution for the production, via dissolution of the specimen, of a sufficient concentration of these ions. Similarly, an increase in the volume of the fresh solution results in an increase in time-to-failure.



Fracture limit diagram illustrating the influence of deformation path on the slow strain rate stress-corrosion cracking of Admiralty brass in $0.1M$ CuSO_4 solution at open circuit potential. The equivalent strain rate is $8 \times 10^{-6}/s$.

The composition of Admiralty brass is 70.9 to 71.8% Cu, 28.2 to 27.1% Zn, 1.0 to 0.8% Sn, and 0.02 to 0.10% As. The brass undergoes transgranular stress-corrosion cracking characterized by cleavage-like cracks propagating on a macroscopic plane normal to the maximum principal strain axis for all strain paths. The average crack growth velocity is also independent of strain path. It was thus concluded that the mechanism of stress-corrosion cracking in this system does not depend on a multiaxial strain path. The lower ductility in equiaxial tension was attributed to stress-corrosion cracks acting as linear imperfections whose presence causes failure of the uncracked ligament by a form of localized necking and whose influence is dependent not only on time, but also on strain path.

Alpha Brass: Effect of Potential on Stress-Corrosion Crack Growth Rate, in Neutral Ammonia

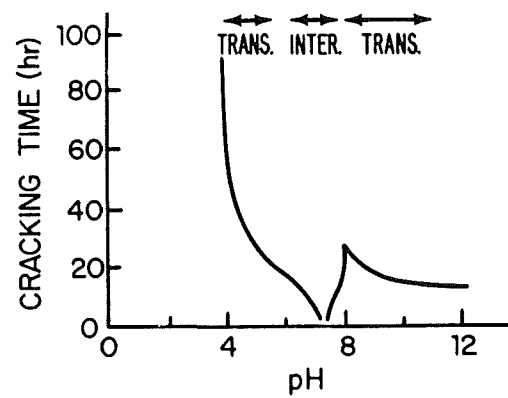


Effect of potential on the growth rate of stress-corrosion cracks in neutral ammonia.

For α -brass, anodic polarization may cause a decrease in the crack growth rate when the potential is moved into the region where the surface film is less stable. In addition, hydrogen may play a role. In many alloys, increasing anodic polarization may generate hydrogen at a higher rate locally in the region of the crack tip.

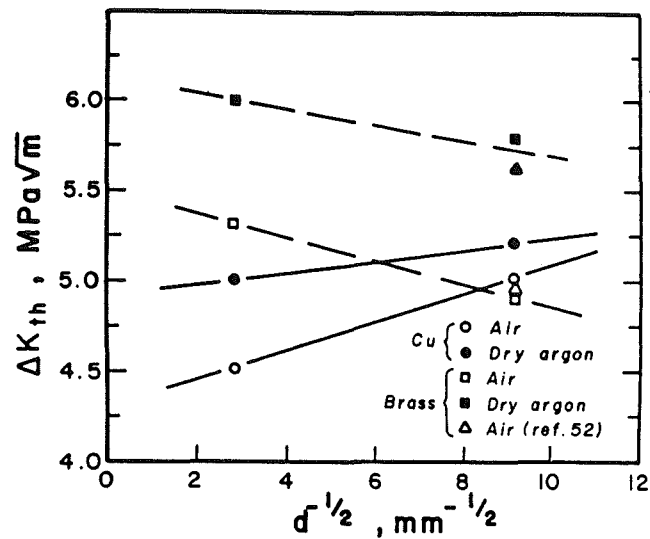
Alpha Brass: Time-to-Cracking, in Ammoniacal Sulfate Solutions

367



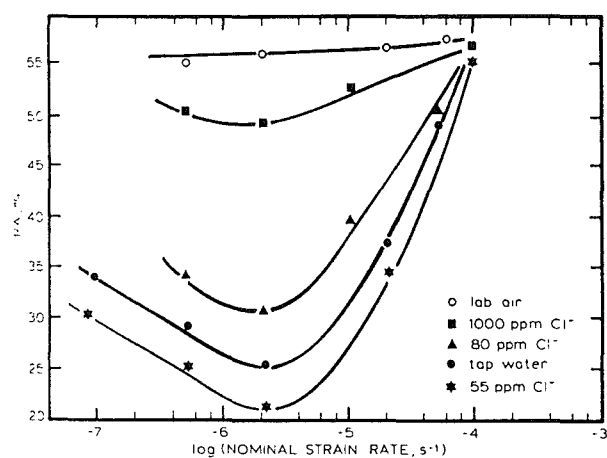
Time-to-cracking of α -brass in ammoniacal sulfate solutions of varying pH.

Copper and Alpha Brass: Influence of Grain Size and Environment on Threshold Value of Stress-Intensity Range



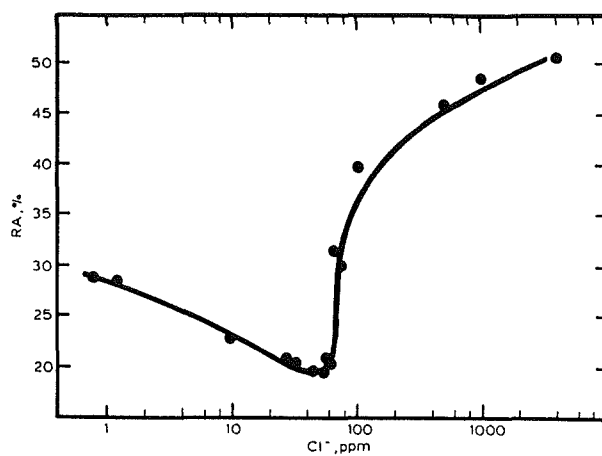
Influence of grain size and environment on the threshold value of the stress-intensity range for copper and α -brass.

The increase in threshold with increasing grain size in α -brass was attributed to the dominance of crack closure effects. In copper, on the other hand, the decrease in threshold with increasing grain size was attributed to the dominance of strength effects.



Variation in ductility of α - β brass with nominal strain rate in water containing various chloride levels.

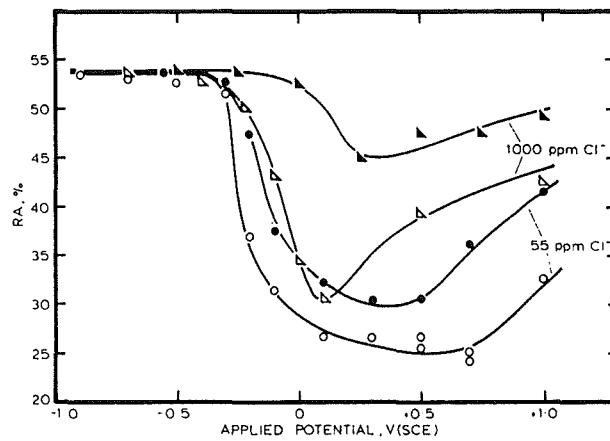
Alpha-Beta Brass: Variation of Ductility with Chloride Content



Variation of ductility of α - β brass with chloride content of testing environment for a nominal strain rate of 2×10^{-6} /s.

Alpha-Beta Brass: Variation of Ductility with Applied Potential for Various Strain Rates

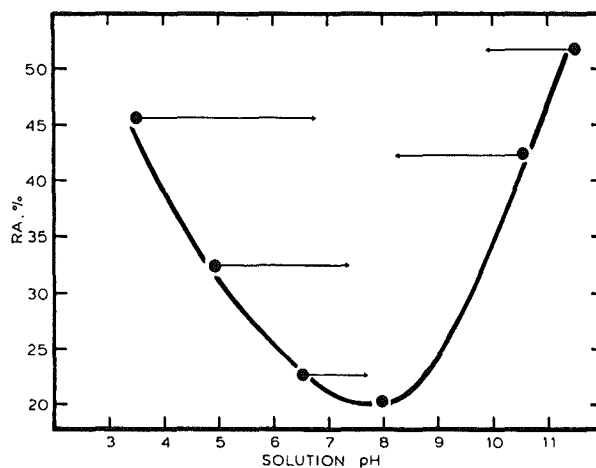
371



Variation of ductility of α - β brass with applied potential for various strain rates.

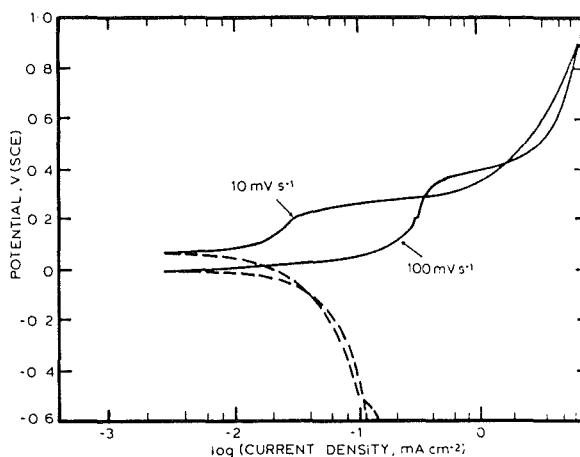
Composition: 38.9-0.29Zn bal Cu. The alloy had been extruded and lightly cold drawn to final size. The microstructure consisted of stringers of a β in an α matrix.

Alpha-Beta Brass: Effect of Initial pH of Environment on Reduction in Area to Fracture



Effect of initial pH of environment on reduction in area to fracture. Arrows indicate the variation of pH during straining.

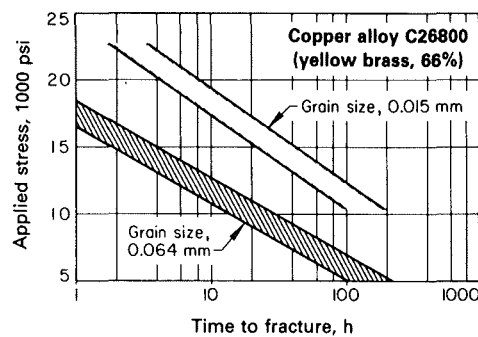
Composition: 38.9-0.29Zn bal Cu. The alloy had been extruded and lightly cold drawn to final size. The microstructure consisted of stringers of a β in an α matrix.



Polarization curves at two sweep rates for α - β brass in water containing 55 ppm Cl^- . Full curve represents anodic polarization, and broken curve represents cathodic polarization.

Composition: 38.9-0.29Zn bal Cu. The results indicated a pitting potential in the region of 290 mV_{SCE}, compared with the free corrosion potential of 0 to 60 mV_{SCE}. Extensive corrosion, with relatively high current densities, was indicated above 300 mV_{SCE}. The cracking range predicted from the significant separation of the fast and slow sweep rate curves was 10 to 250 mV_{SCE}. The observed incidence of stress-corrosion cracking was centered around this potential region, but covered a more extensive range of potentials.

Copper Alloy C26800: Effect of Grain Size on Time-to-Fracture, in Ammonia Atmosphere

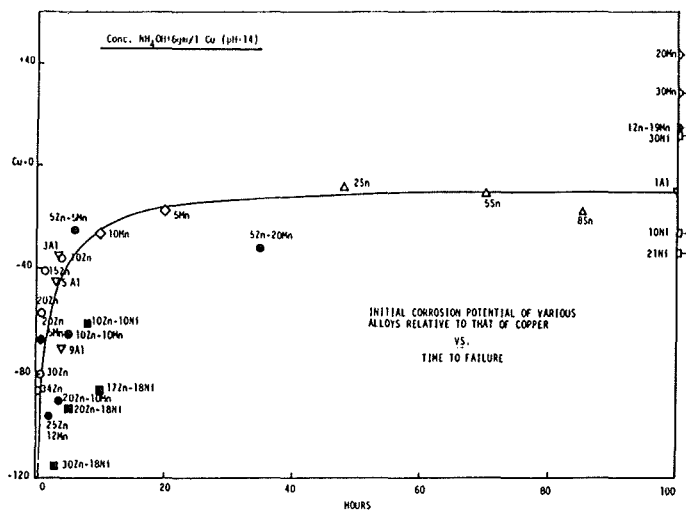
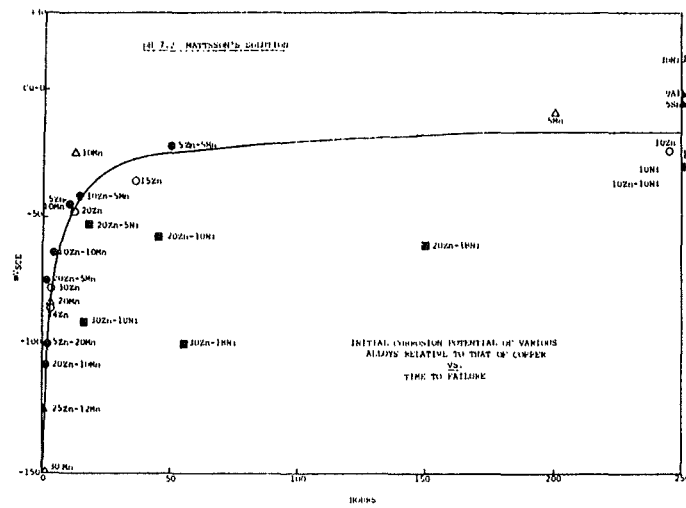


Effect of grain size on the time-to-fracture of copper alloy C26800 (yellow brass, 66%) in ammonia atmosphere.

In general, any metal with a small grain size is more resistant to stress-corrosion cracking than the same metal having a large grain size. This relationship applies whether crack propagation is intergranular or transgranular.

Copper Alloys: Initial Corrosion Potential Versus Stress-Corrosion Cracking Failure

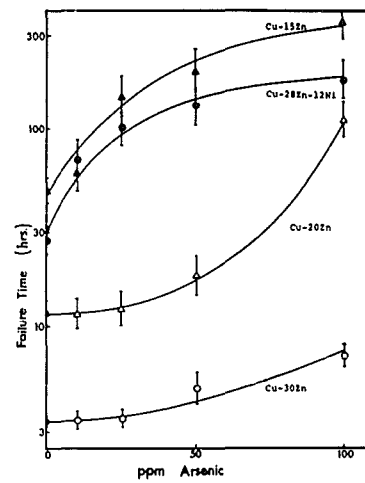
375



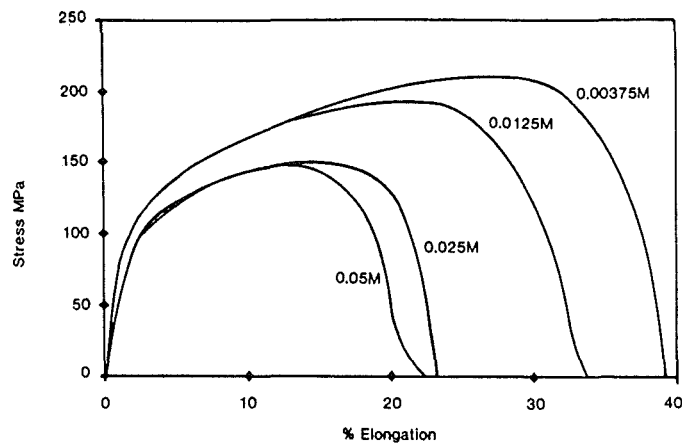
Initial corrosion potential (relative to Cu) versus stress-corrosion cracking failure time for various alloys in pH 7.2 solution.

Source: A. Parthasarathi and N.W. Polan, Stress Corrosion Cracking of Copper Alloys: The Effects of Corrosion Potential and Tarnishing Characteristics, Corrosion, 43(12), December 1987, p 747-755.

Copper Alloys: Effect of Arsenic on Stress-Corrosion Cracking

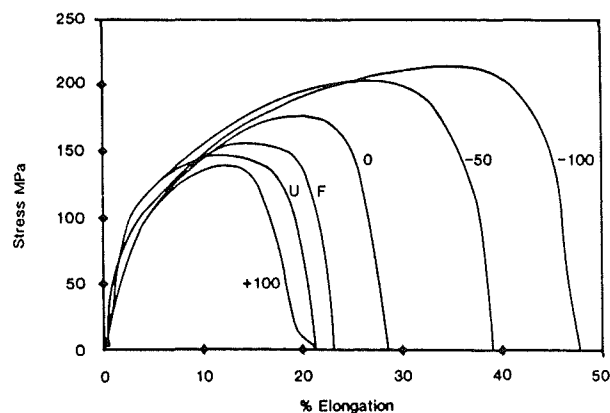


Effect of adding arsenic to pH 7.2 Mattson's solution on stress-corrosion cracking failure for several copper alloys.



Stress-elongation curves obtained from slow strain rate testing of oxygen-free copper in aerated solutions of NaNO_2 at the concentrations shown at room temperature and at an electrode potential of $+100 \text{ mV}_{\text{SCE}}$; strain rate $2.6 \times 10^{-6}/\text{s}$.

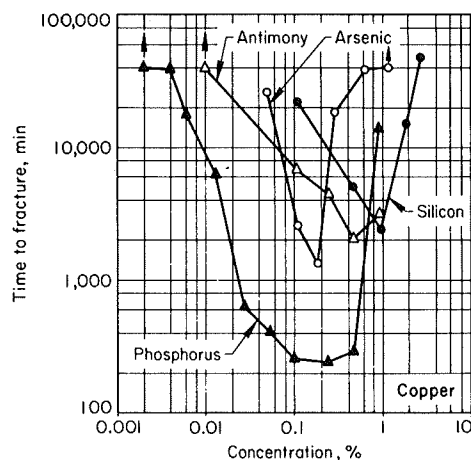
Nitrite-induced stress-corrosion cracking in copper water pipes has been associated with residual stresses introduced during manufacture or installation. Nitrite is a well-known corrosion inhibitor for steel and had been added to heat insulation materials applied to the outside of the pipe where cracking initiated. With the elimination of nitrite, the incidence of stress-corrosion cracking decreased dramatically.

Copper (Cu-OF): Stress-Elongation Curves, in NaNO_2 

Stress-elongation curves obtained from slow strain rate testing of oxygen-free copper in aerated $0.6M$ NaNO_2 solution at room temperature. The figures on the curves indicate applied potential in mV_{SCE} . F = free corrosion potential and U = +50 mV; the strain rate was $2.6 \times 10^{-6}/\text{s}$.

Copper: Effect of Elemental Additions on Stress-Corrosion Cracking, in Ammoniacal Atmosphere

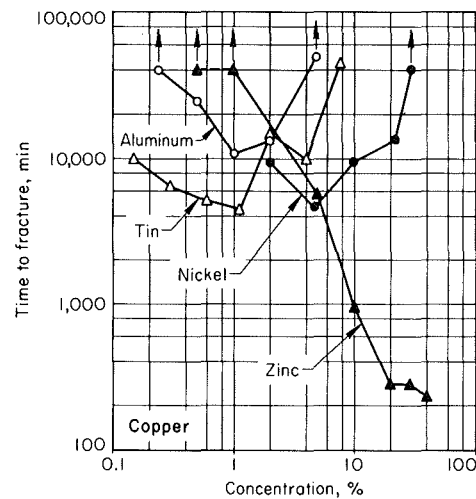
379



Effects of arsenic, phosphorus, antimony, and silicon on stress-corrosion cracking of copper.

The effect of low concentrations of arsenic, phosphorus, antimony, and silicon on time-to-fracture of copper by stress-corrosion cracking under an applied tensile stress of 69 MPa in a moist ammoniacal atmosphere is indicated. The composition of the test atmosphere was 80% air, 16% ammonia, and 4% water vapor. The temperature was 35 °C, which was above the dew point.

Copper: Effect of Elemental Additions on Stress-Corrosion Cracking, in Ammoniacal Atmosphere

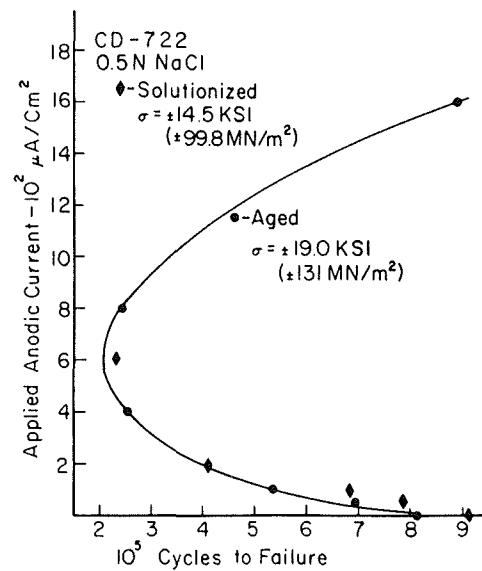


Effects of aluminum, nickel, tin, and zinc on stress-corrosion cracking of copper.

The effect of low-to-moderate concentrations of aluminum, nickel, tin, and zinc on the time-to-fracture of copper by stress-corrosion cracking under an applied tensile stress of 69 MPa in a moist ammoniacal atmosphere is indicated. The composition of the test atmosphere was 80% air, 16% ammonia, and 4% water vapor. The temperature was 35 °C, which was above the dew point.

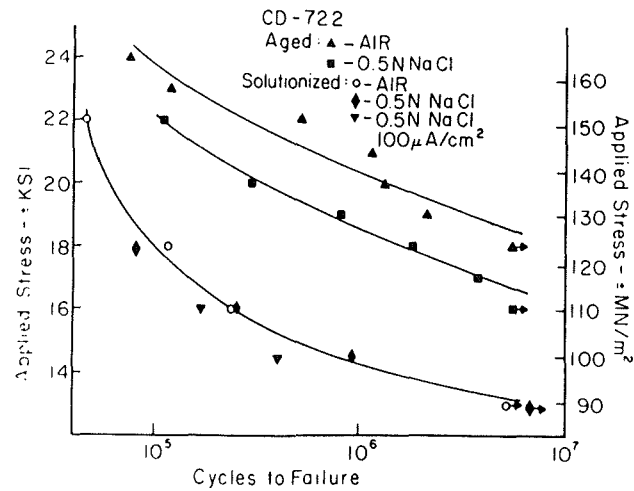
Copper Alloy IN838 (CDA 722): Effect of Applied Anodic Currents on Fatigue Resistance, in NaCl Solution

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Effect of applied anodic currents on the fatigue resistance of solutionized and age hardened IN838 in 0.5N NaCl solution at 20 Hz.

Copper Alloy IN838 (CDA 722): Stress Versus Number of Cycles-to-Failure, in Laboratory Air and in NaCl Solution

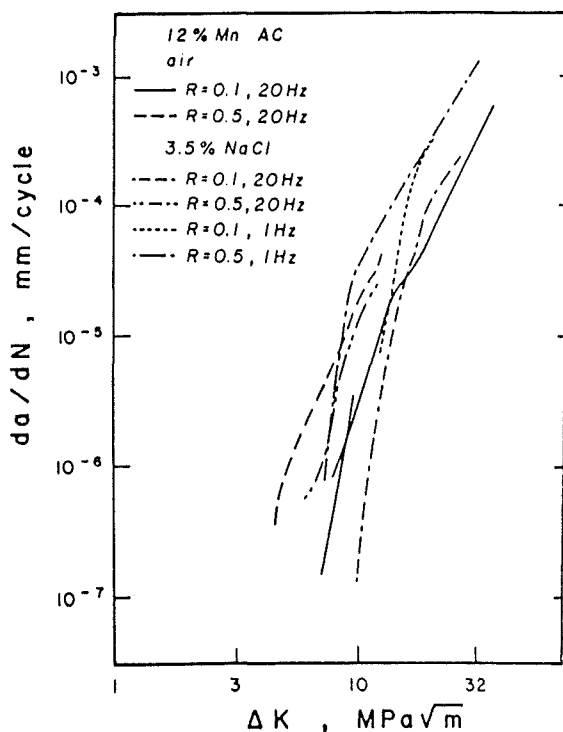


Stress versus number of cycles-to-failure for solutionized and age hardened Cu-Ni-Cr alloy, IN838 (CDA 722), in laboratory air and in aerated 0.5N NaCl solution under free corrosion conditions at 20 Hz.

The composition of IN838 is 16.1% Ni, 0.55% Cr, 0.88% Fe, 0.68% Mn, 0.032% Si, 0.045% Ti, and bal Cu.

12% Mn AC Alloy: Fatigue Crack Growth Rate, in Air and in NaCl Solution

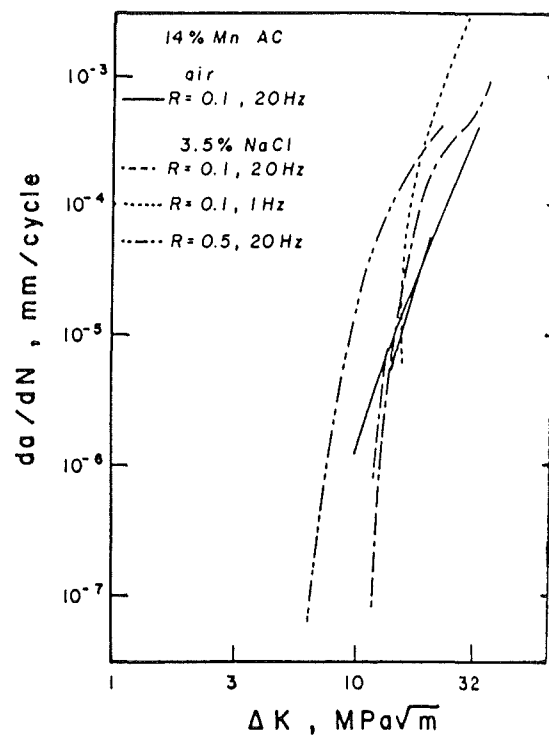
383



Rate of fatigue crack growth for 12% Mn AC alloy in air and in 3.5% NaCl solution.

Typical composition: 2.1Ni-7.9Al-3.3Fe-12Mn bal Cu. Alloys of this type are of interest as ship propeller bronzes.

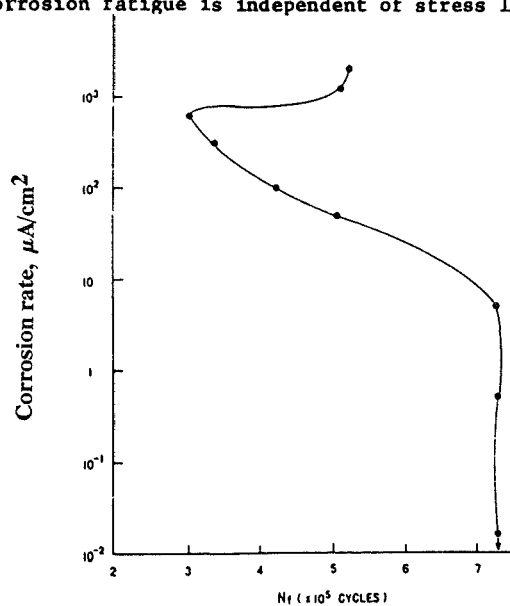
14% Mn AC Alloy: Fatigue Crack Growth Rate, in Air and in NaCl Solution



Rate of fatigue crack growth for 14% Mn AC alloy in air and in 3.5% NaCl solution.

Typical composition: 2.1Ni-7.9Al-3.3Fe-14Mn bal Cu.

corrosion fatigue is independent of stress level.



Effect of applied current on the fatigue resistance of polycrystalline copper.

A critical corrosion rate of 1 to 10 $\mu\text{A}/\text{cm}^2$ is needed to induce corrosion fatigue. The longer lives at very high corrosion rates are associated with the dissolution of nascent cracks.

NICKEL-BASE ALLOYS

NICKEL-BASE ALLOYS

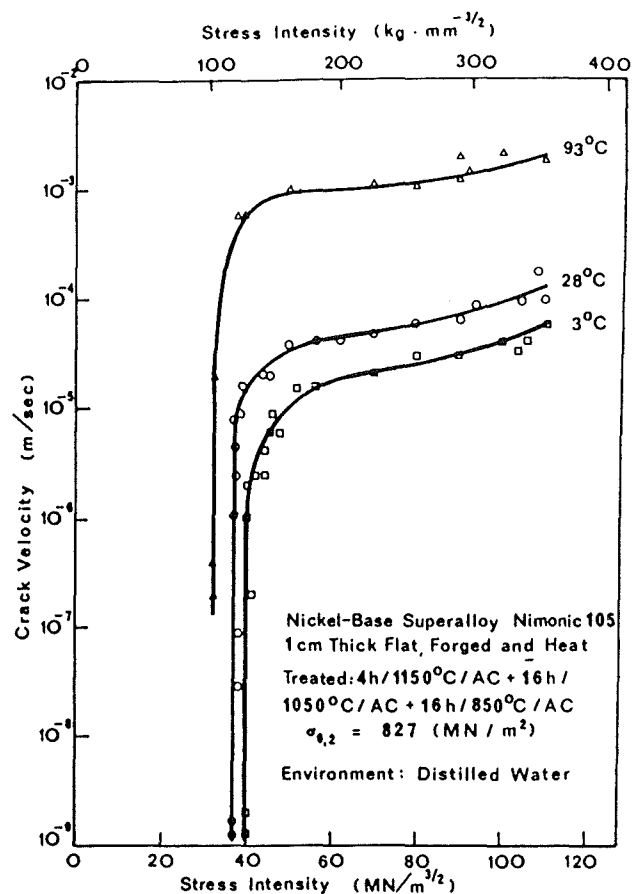
Nickel-base alloys are generally used to combat stress-corrosion cracking where austenitic stainless steels have failed because of stress-corrosion cracking (SCC). However, two events have recently occurred that require increased knowledge of the SCC resistance of nickel-base alloys. First, a large number of alloys have been developed and included in the market; this has resulted in an almost continuous change in performance (alloy content) between stainless steels and the numerous nickel-base alloys. Secondly, historically nickel-base alloys have been considered to be immune to SCC in all but a few environments, but the increased requirements for current practices have extended the use of materials to temperatures at which the SCC of nickel-base alloys must now be considered.

SCC of nickel-base alloys has been found to occur in three types of environments: high-temperature halogen ion solutions, high-temperature water, and high-temperature alkaline environments. In addition, SCC has been detected in liquid metals, near-ambient temperature polythionic acid solutions, and environments containing acids and hydrogen sulfide (H_2S).

With respect to corrosion fatigue (CF) of nickel-base alloys, elevated temperature oxidation, and corrosion associated with multi-component gas mixtures and liquid metals are areas of concern.

Nimonic 105: Effect of Stress-Intensity Factor and Temperature on Stress-Corrosion Cracking, in Water

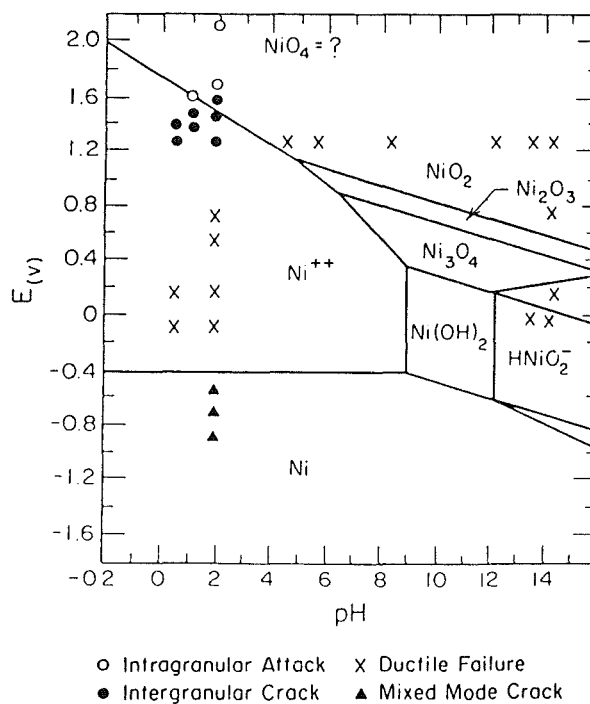
391



Effect of stress-intensity factor and temperature on stress-corrosion crack velocity in Nimonic 105 exposed to water.

Composition: 15% Cr, 20% Co, 5% Mo, 4.5% Al, 1.4% Ti, 0.15% C, bal Ni.

Nickel Alloy 200: Fracture Mode Determined in Constant Extension Rate Tests

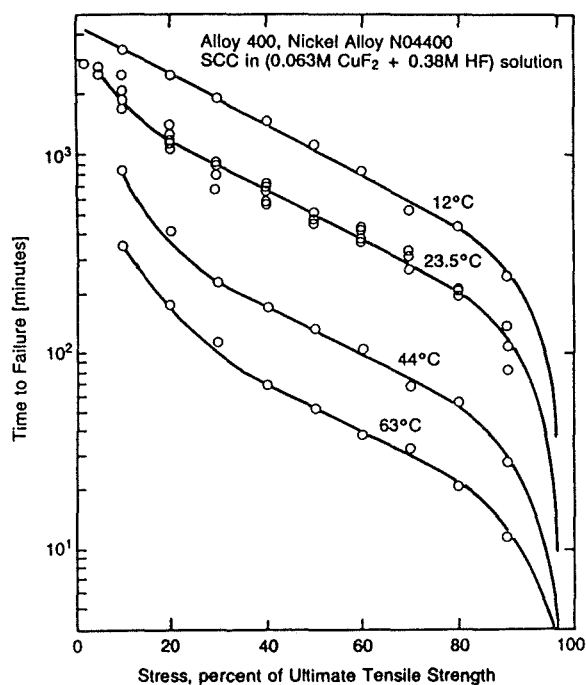


Fracture mode determined in constant extension rate tests of Nickel 200, a commercial nickel containing impurities, superimposed onto a Pourbaix diagram.

Composition: 0.15% Fe, 0.005% S, 0.1% C, 0.1% Ca, 0.05% Si, 0.2% Mn, bal Ni. Sulfur was found to be the main segregating species at thermal exposures in the range 550 to 850 °C. Segregated sulfur was shown to catalyze the anodic dissolution kinetics when nickel is exposed to sulfuric acid.

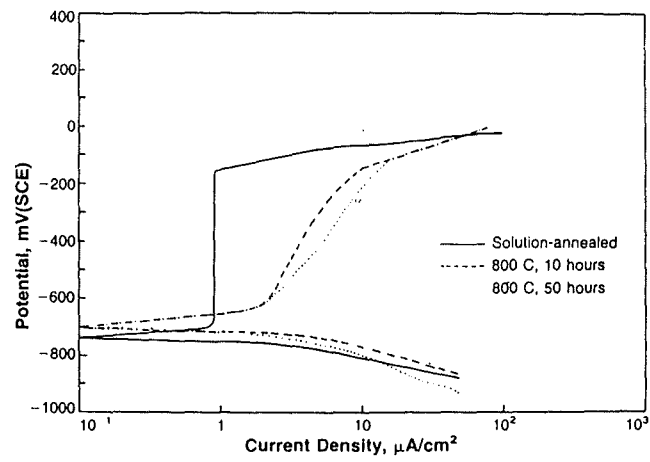
Nickel Alloy 400: Effect of Stress and Temperature on Stress-Corrosion Cracking Resistance, in HF + CuF₂ Solutions

393

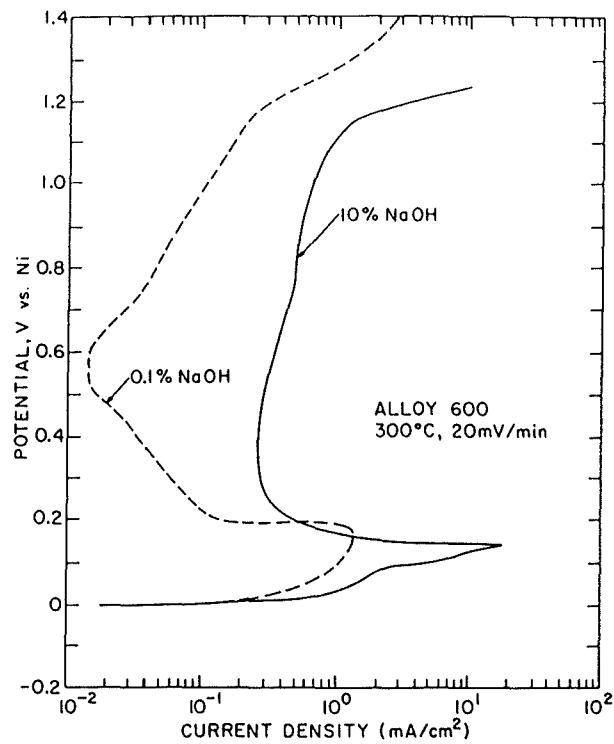


Effect of stress and temperature on stress-corrosion cracking resistance of Alloy 400 in HF + CuF₂ solutions.

Composition: 0.3% max C, 25 to 32% Cu, 63% Ni. In this environment, stress-corrosion cracking can occur in Alloy 400 at relatively low stress levels.

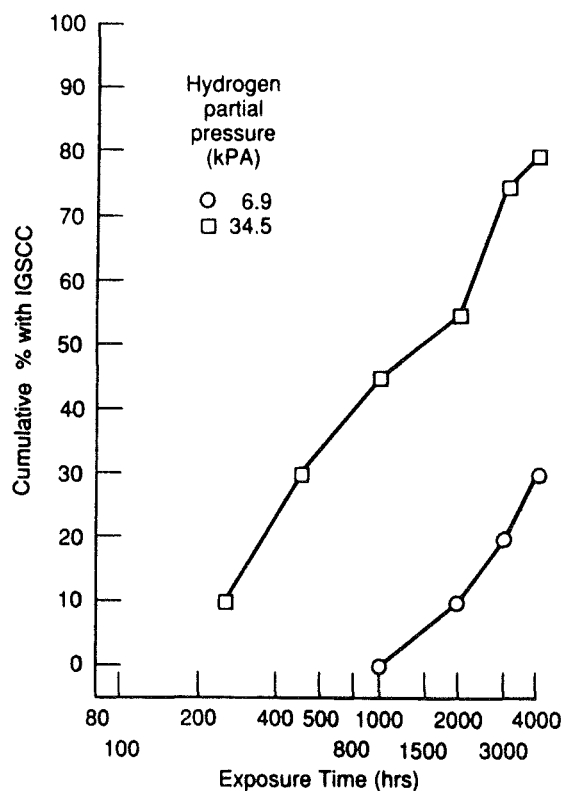
Nickel Alloy 600: Polarization Curves, in NaCl Solution

Polarization curves for solution annealed and 800 °C heat treated Alloy 600 in deaerated 4M NaCl solution.



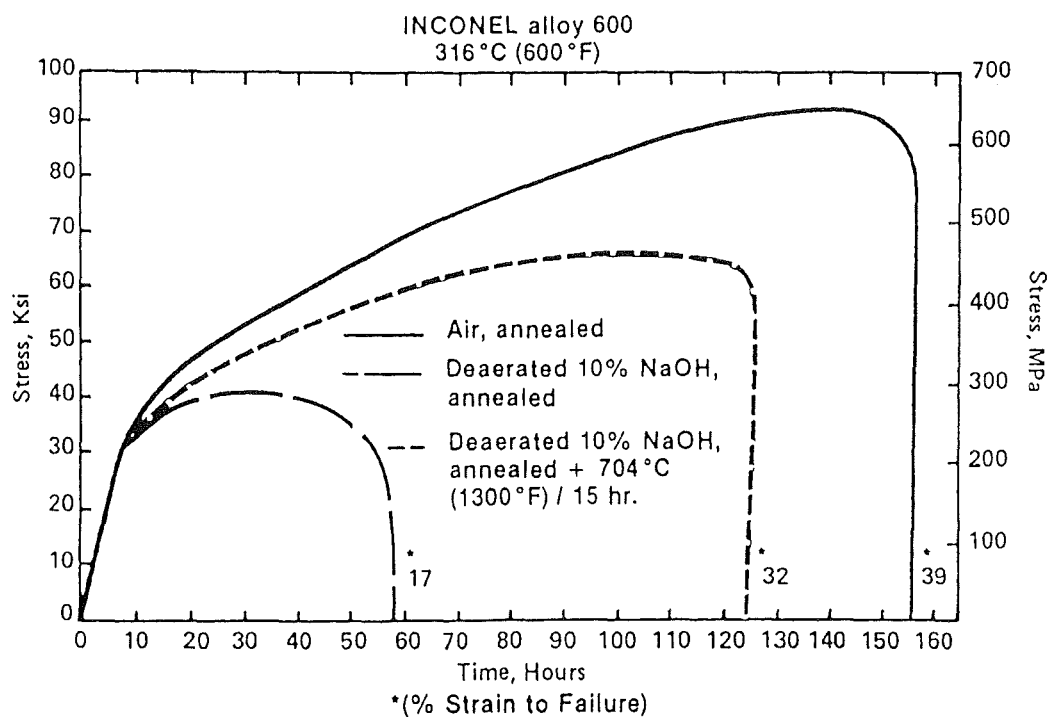
Polarization curves for Alloy 600 in 10% and in 0.1% NaOH at 300 °C.

Nickel Alloy 600: Effect of Hydrogen Partial Pressure on the Time to Induce Intergranular Stress-Corrosion Cracking, in Pure Water



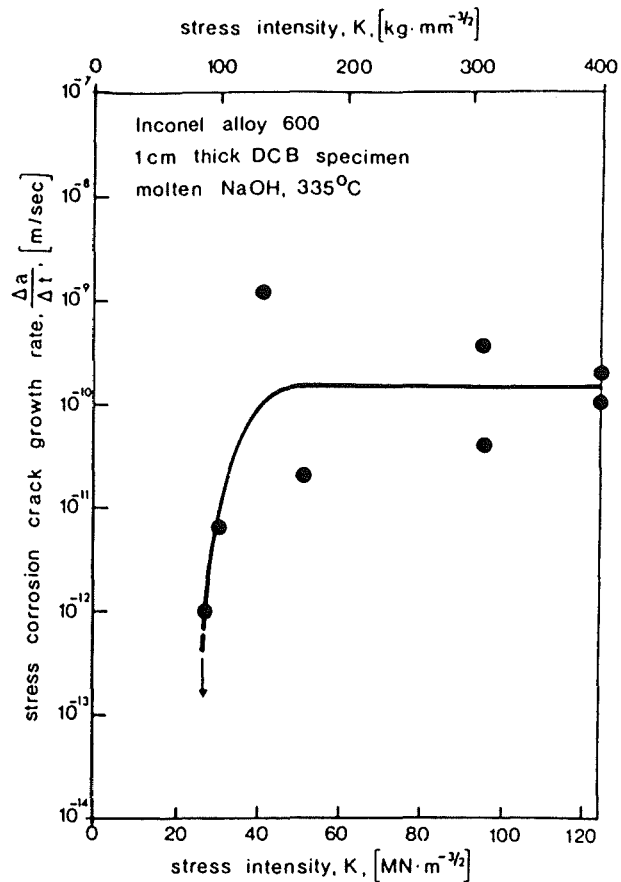
Effect of hydrogen partial pressure on the time to induce intergranular stress-corrosion cracking in mill annealed Alloy 600 specimens in pure water at 360 °C.

Source: R.S. Pathania and A.R. McIlree, A Review of the Effect of Hydrogen on Stress Corrosion Cracking of Alloy 600 in 360 °C Water, in Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors, G.J. Theus and J.R. Weeks, Ed., The Metallurgical Society of AIME, Warrendale, PA, 1988, p 551-554.



Controlled extension rate test results for Alloy 600. Strain rate, 1×10^{-6} /s; temperature, 316 °C.

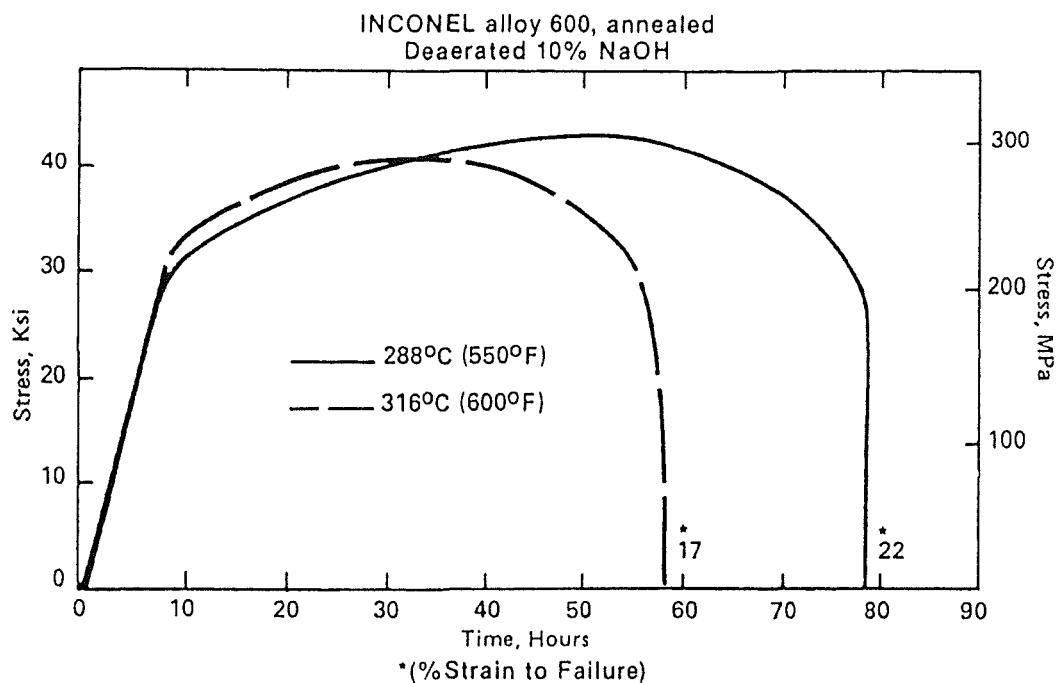
Nickel Alloy 600: Stress-Corrosion Crack Velocity as a Function of the Stress-Intensity Factor, in Molten NaOH



Stress-corrosion crack velocity in sensitized Alloy 600 exposed to molten NaOH at 335°C as a function of the stress-intensity factor.

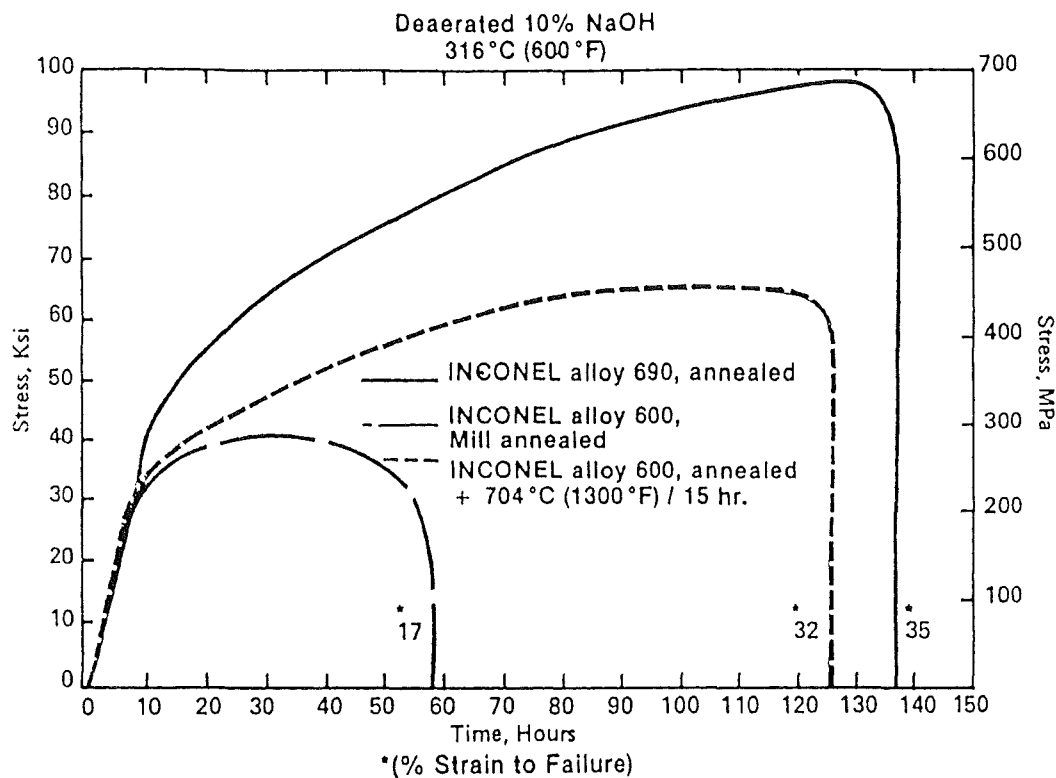
Nickel Alloy 600: Controlled Extension Rate Test Results, in NaOH

399

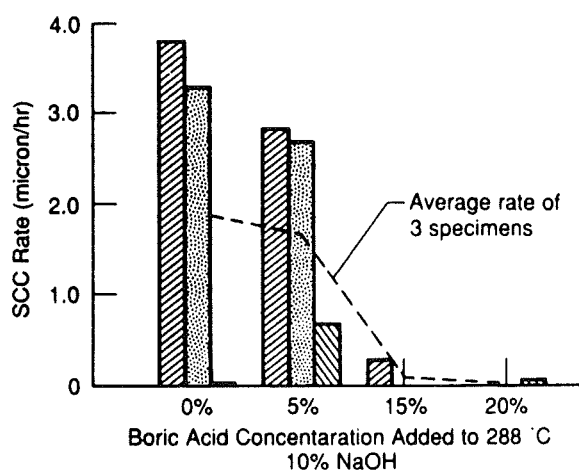


Controlled extension rate test results for annealed Alloy 600 in deaerated 10% NaOH and at 150 mV versus Pt; strain rate, 1×10^{-6} /s.

Nickel Alloys 600 and 690: Controlled Extension Rate Test Results, in NaOH



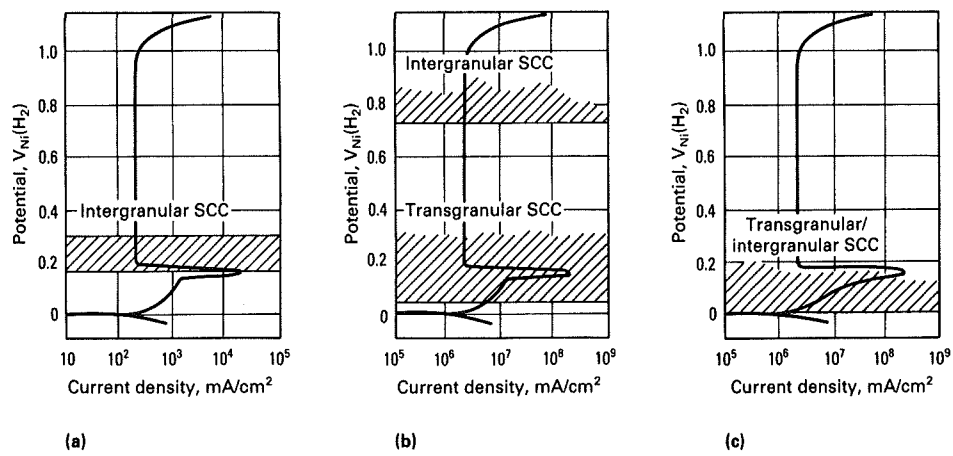
Controlled extension rate test results for Alloys 600 and 690 in deaerated 10% NaOH and at 150 mV versus Pt; strain rate, 1×10^{-6} /s; temperature, 316 °C.



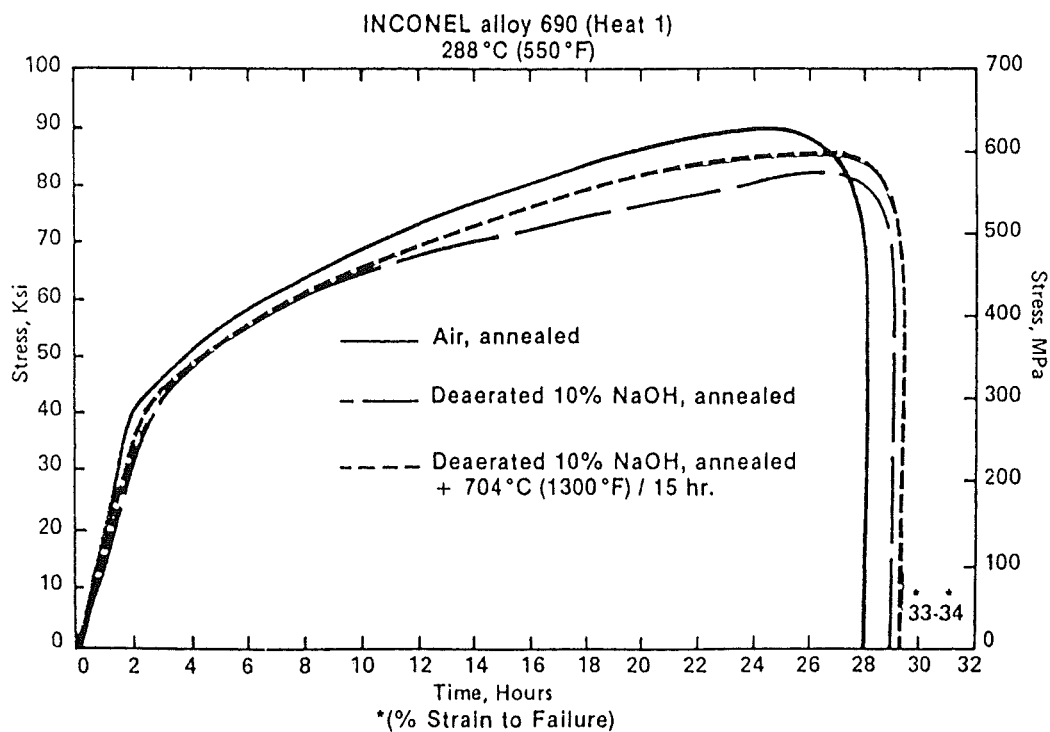
Effect of boric acid on stress-corrosion cracking of alloy 600 in caustic.

The results showed that, in general, boric acid inhibited intergranular stress-corrosion crack growth in alkaline environments. The degree of effectiveness depended on the configuration of the specimen and the method of boric acid application. Access of boric acid to the crack tip was necessary for effective inhibition. A growing number of plants are using boric acid to control intergranular attack and intergranular stress-corrosion cracking.

Nickel Alloys 600 and 800: Intergranular and Transgranular Stress-Corrosion Cracking, in NaOH



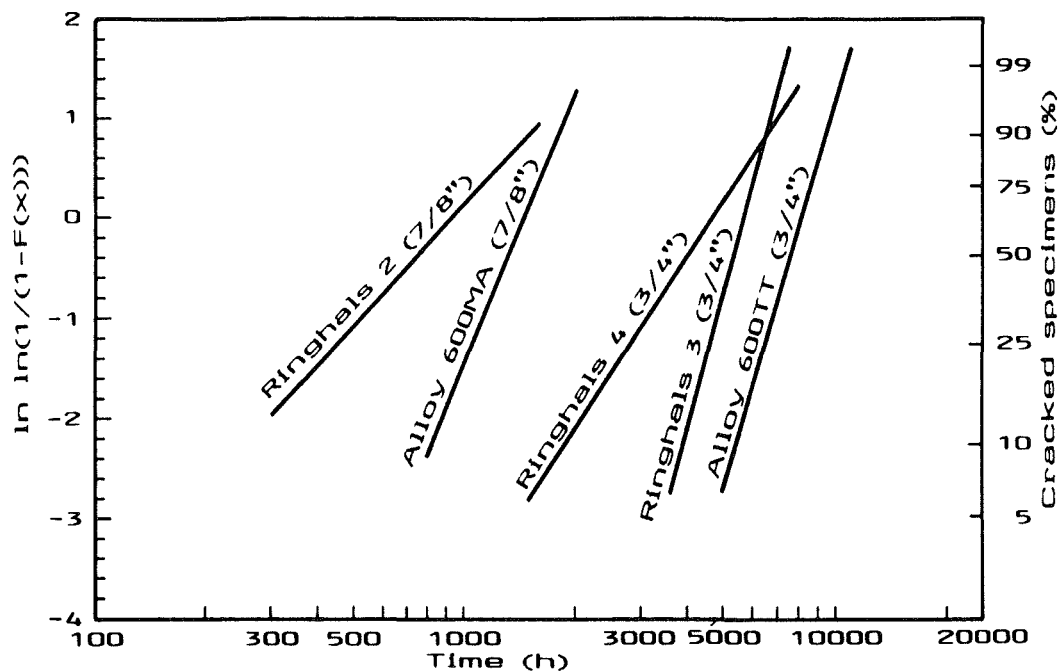
Potentiokinetic polarization curve and electrode potential values at which intergranular and transgranular stress-corrosion cracking appears in a 10% NaOH solution at 288 °C. (a) Alloy 600. (b) Alloy 800. (c) AISI 304 stainless steel.



Controlled extension rate test results for Alloy 690; strain rate, 4.1×10^{-6} /s; temperature, 288 °C.

Under the test conditions, Alloy 690 exhibited a much greater resistance to stress-corrosion cracking than Alloy 600. The effects of heat treatment were less evident in Alloy 690 than in Alloy 600.

Nickel Alloy 600 Tube: Weibull Plots

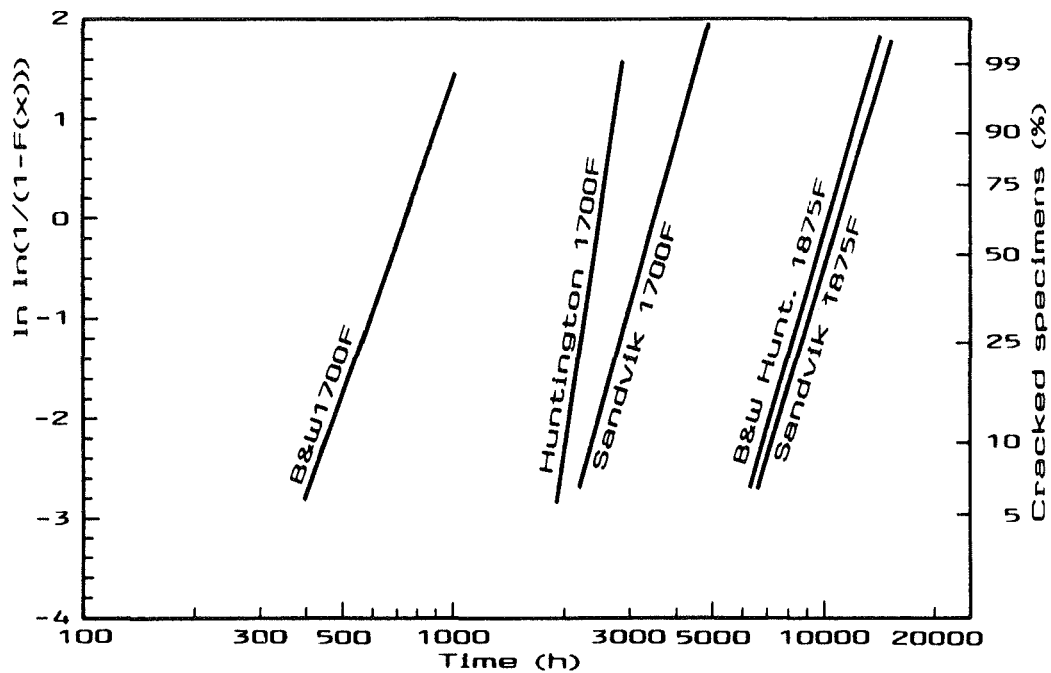


Weibull plots for Alloy 600 tube.

Composition: 0.014% C, 0.36% Si, 0.70% Mn, 0.004% S, 0.012% P, 16.07% Cr, 72.75% Ni, 0.030% Co, 0.18% Ti, 0.14% Cu, 0.12% Al, 0.030% Mo, 0.006% N, 9.34% Fe, 0.0016% B. The tests were performed at 365 °C with reverse U-bend specimens in high-purity water with hydrogen added.

Source: K. Norring, Intergranular Stress Corrosion Cracking in Steam Generator Tubing. Testing of Alloy 690 and Alloy 600 Tubes, in Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors, G.J. Theus and J.R. Weeks, Ed., The Metallurgical Society of AIME, Warrendale, PA, 1988, p 587-593.

Source: K. Norring, Intergranular Stress Corrosion Cracking in Steam Generator Tubing. Testing of Alloy 690 and Alloy 600 Tubes, in Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors, G.J. Theus and J.R. Weeks, Ed., The Metallurgical Society of AIME, Warrendale, PA, 1988, p 587-593.

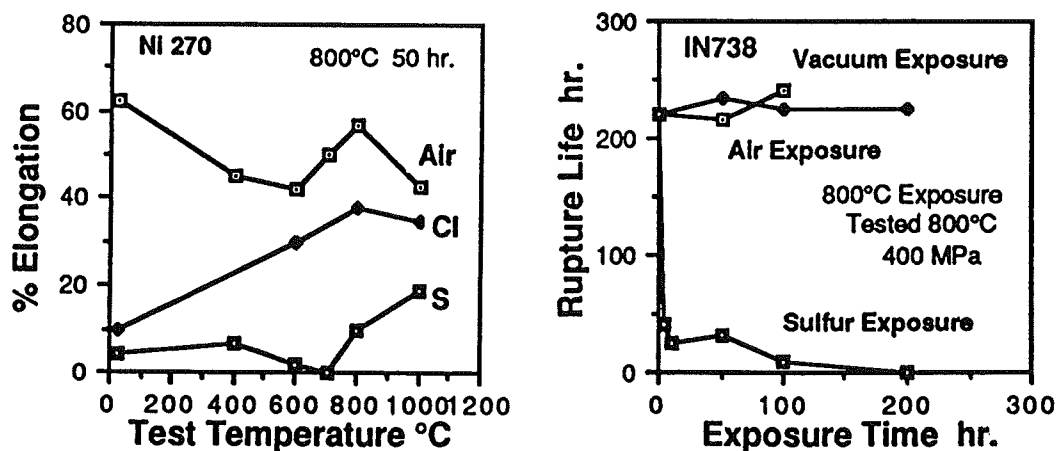


Weibull plots for EPRI special production tube.

The tests were performed at 365 °C with reverse U-bend specimens in high-purity water with hydrogen added.

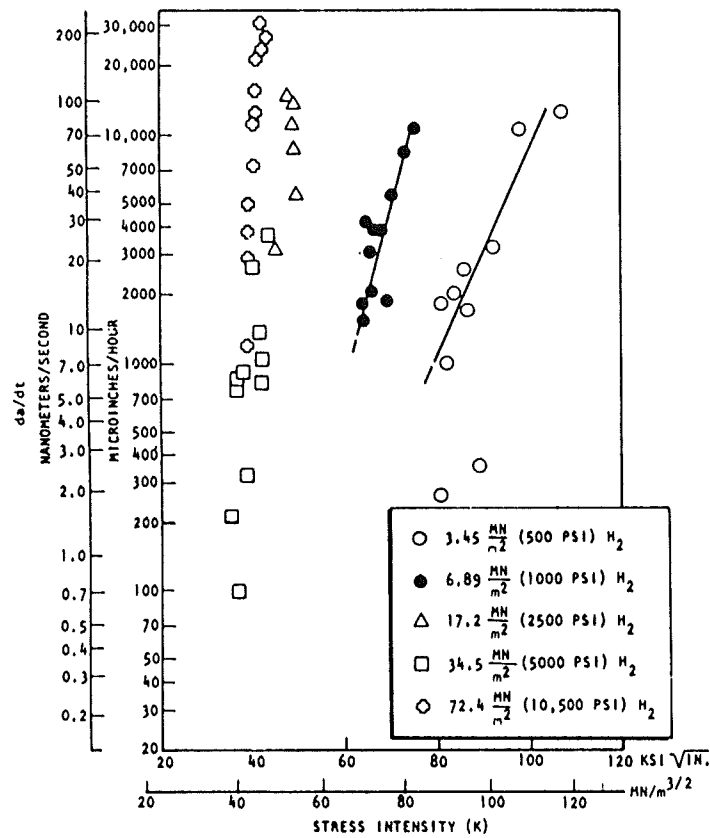
Source: K. Norring, Intergranular Stress Corrosion Cracking in Steam Generator Tubing. Testing of Alloy 690 and Alloy 600 Tubes, in Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors, G.J. Theus and J.R. Weeks, Ed., The Metallurgical Society of AIME, Warrendale, PA, 1988, p 587-593.

IN738: Effect of Prior Exposure on Stress-Rupture Properties, in Air, Vacuum, and in Sulfur



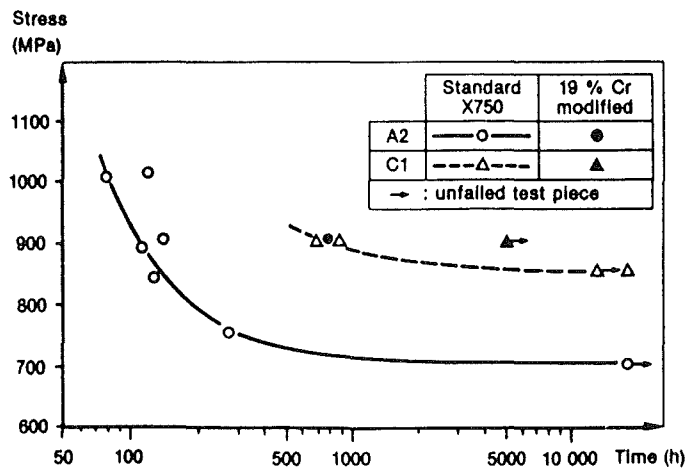
Effect of prior 800 °C exposure in air, vacuum, and sulfur on the 800 °C stress-rupture properties on IN738.

The diffusion of oxygen from the environment down grain boundaries, resulting in embrittlement and enhanced crack propagation, has been documented in nickel and nickel-base superalloys. (D.A. Woodford and R.H. Bricknell, in *Treatise on Materials Science and Technology*, Vol 25, C.L. Briant and S.K. Banerjee, Ed., Academic Press, New York, p 157-199). Work on the effects of grain boundary penetration of sulfur in nickel and nickel-base superalloys indicates that the tensile ductility, stress-rupture life, and crack propagation resistance are all more seriously impaired by sulfur than by oxygen.



Sustained load crack growth rate in Inconel 718 as a function of stress-intensity factor at various hydrogen pressures.

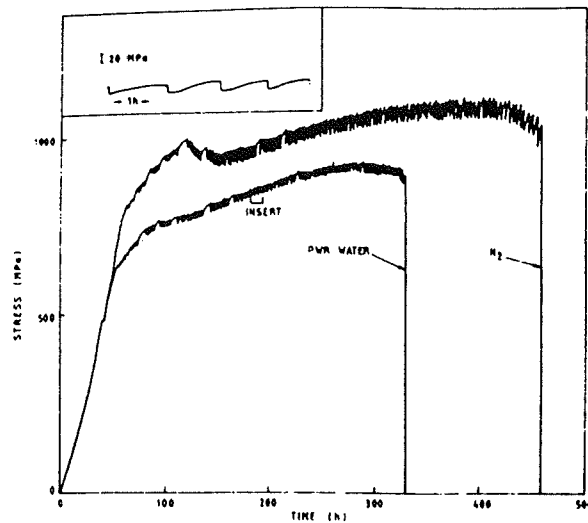
Standard and Modified Nickel Alloy X750: Stress-Corrosion Cracking Tests, in Pure Water



Stress-corrosion cracking tests on tensile specimens of standard and modified Alloy X750 in pure water with 25 to 50 ml H₂ (STP) per kg H₂O at 350 °C.

Composition: 0.063% C, 0.22% Si, 0.05% Mn, 0.006% S, 16.22% Cr, 71.80% Ni, 0.04% Co, 2.57% Ti, 0.015% Cu, 0.54% Al, 0.12% Mo, 1.00% Ta + Nb, 7.54% Fe. In the modified alloy, the chromium content has been increased to 19% as compared to 16.22% for Alloy X750. The microstructure and mechanical properties of the modified alloy are similar to those of the standard alloy. However, the stress-corrosion cracking resistance of the modified alloy in deaerated hydrogenated primary or pure water at 350 to 360 °C and in hydrogenated steam at 400 °C is appreciably higher.

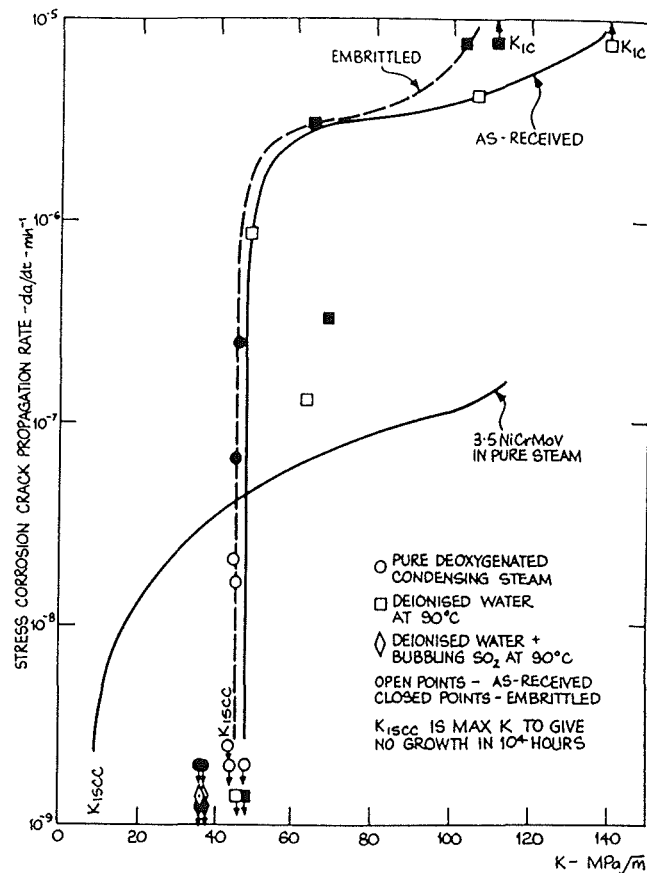
Source: J.R. Donati, et al., Stress Corrosion Cracking Behavior of Nickel-Base Alloys with 19% Chromium in High Temperature Water, in Environmental Degradation of Materials in Nuclear Power Systems—Water Reactors, G.J. Theus and J.R. Weeks, Ed., The Metallurgical Society of AIME, Warrendale, PA, 1988, p 697-701.



Stress-time curves for Inconel X750 (980 °C for 1 h + 885 °C for 24 h + 704 °C for 20 h) tested at 340 °C in N₂ and in PWR primary water at 2.0×10^{-7} /s.

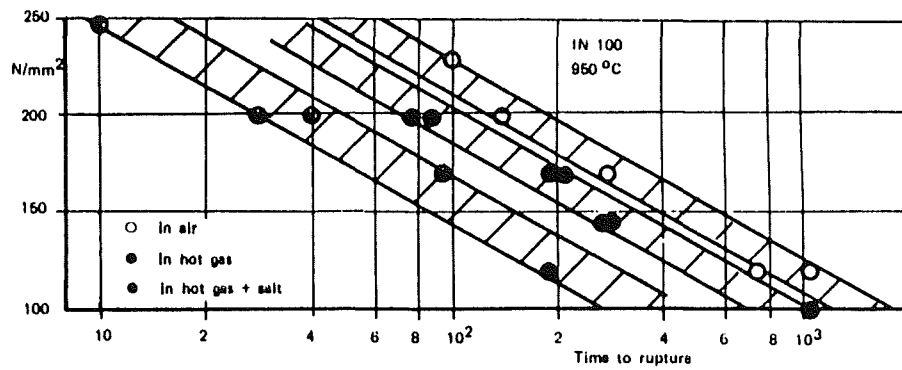
At the corrosion potential (-670 mV, Ag/AgCl), in low (<2 ppb) oxygen concentrations, a high-temperature (1093 to 1107 °C) solution anneal and a single-stage aging treatment has been shown to give improved stress-corrosion cracking resistance as compared with a 980 °C solution anneal or other two-stage aging treatments. However, material given the first heat treatment can be susceptible to cracking at imposed potentials of greater than -200 mV (Ag/AgCl) if the alloy is severely sensitized.

Nickel Alloy 80A: Rate Comparison of Stress-Corrosion Crack Growth as a Function of K



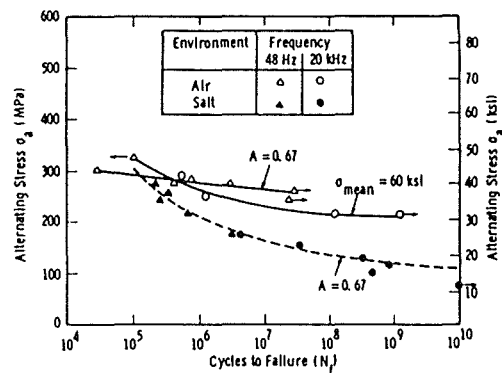
Comparison of the rate of stress-corrosion crack growth as a function of K in Alloy 80A with that in 3.5NiCrMoV.

Alloy 80A (Ni-20Cr-2.4Ti-1.4Al) is used in steam turbines as a bolting material and experiences temperatures in the 450 to 565 °C range. The material is used because of its high creep strength and stress relaxation resistance, which permits the use of smaller bolts and more compact flanges than is possible with ferritic bolting materials. However, approximately 0.2% of the bolts in service have failed because of stress-corrosion cracking.



Effect of corrosion on the time-to-rupture on IN 100 at 950 °C.

Udimet 720: Corrosion Fatigue Behavior, in Air and in Molten Salt

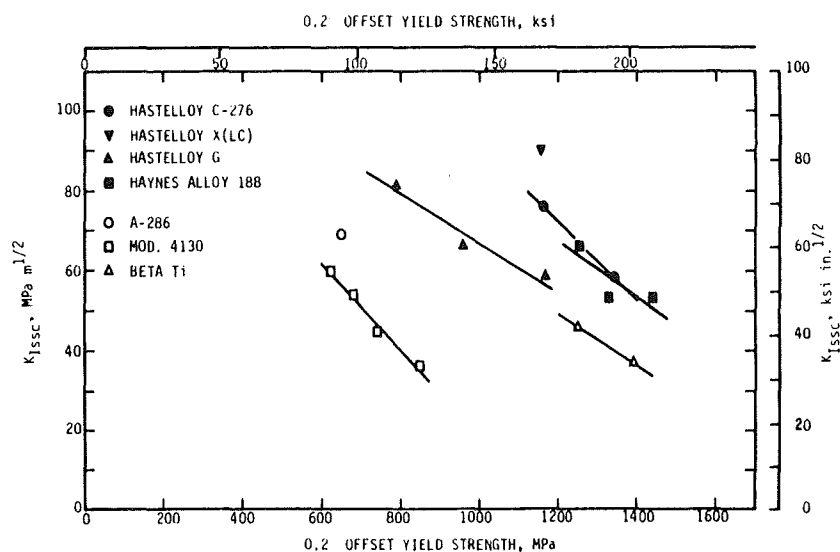


Corrosion fatigue behavior of Udimet 720 at 704 °C in air and in a molten salt environment at test frequencies of 48 Hz and 20 kHz.

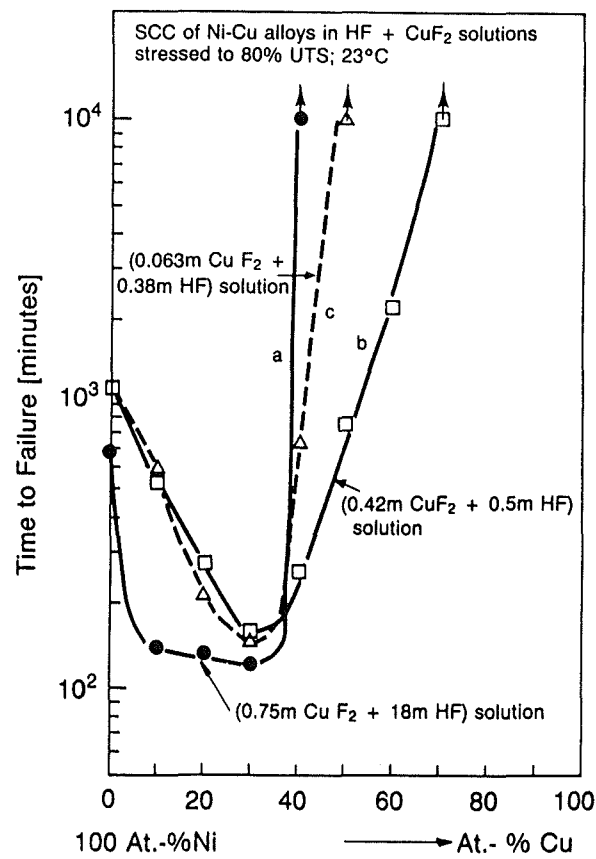
Composition: 54.6% Ni, 18% Cr, 14.8% Co, 4.94% Ti, 3.1% Mo, 2.54% Al, 1.29% W, 0.1% Fe, 0.26% Zr, 0.035% B, 0.003% S, 0.030% C. A is the ratio of the stress amplitude to the mean stress. The ratio of the minimum stress to the maximum stress, R, is more commonly used in characterizing the nature of a loading cycle. The observed degradation in life for the specimens tested in salt suggests that the fatigue resistance is reduced on a per-cycle basis regardless of frequency.

Nickel Alloys: Variation of Fracture Toughness with Yield Strength

413



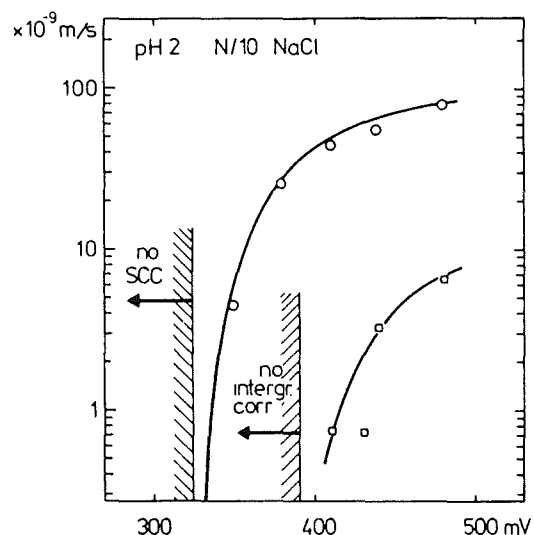
Variation of fracture toughness, K_{ISSC} , with yield strength for materials exposed to NACE aqueous solution containing 5% NaCl, 0.5% CH_3COOH , and saturated with H_2S .



Time-to-failure of stressed nickel-copper alloys in HF + CuF₂ solutions.

Ni-10Pd Nickel Alloy: Propagation Rates of Stress-Corrosion Cracks and Intergranular Corrosion as a Function of Electrode Potential

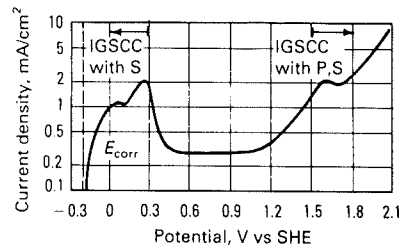
415



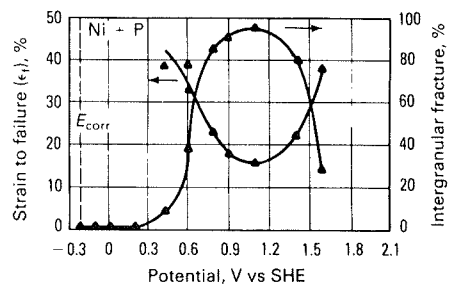
Propagation rates of stress-corrosion cracks (circles) and intergranular corrosion (squares) of Ni-10Pd as a function of electrode potential.

Source: H. Kaiser, E. Lenz, and H. Kaesche, Localized Corrosion and Stress Corrosion of NiCoPd Alloys in Chloride Solutions, in Mechanisms of Environment Sensitive Cracking of Materials, P.R. Swann, F.P. Ford, and A.R.C. Westwood, Ed., The Metals Society, London, 1977, p 92-101.

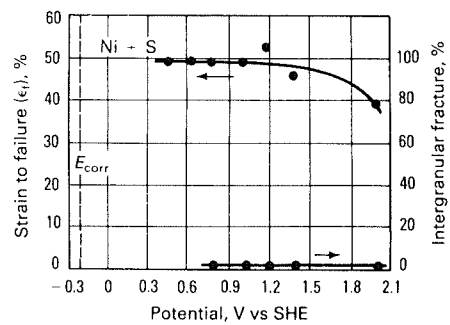
Ni + S and Ni + P Alloys: Stress-Corrosion Cracking Behavior, in H₂SO₄



(a)

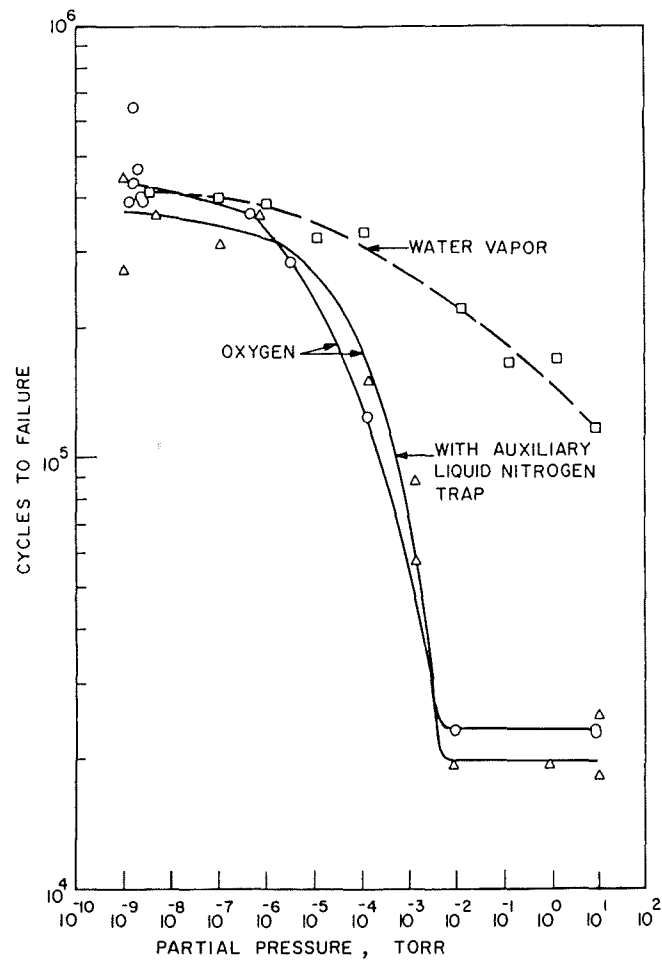


(b)



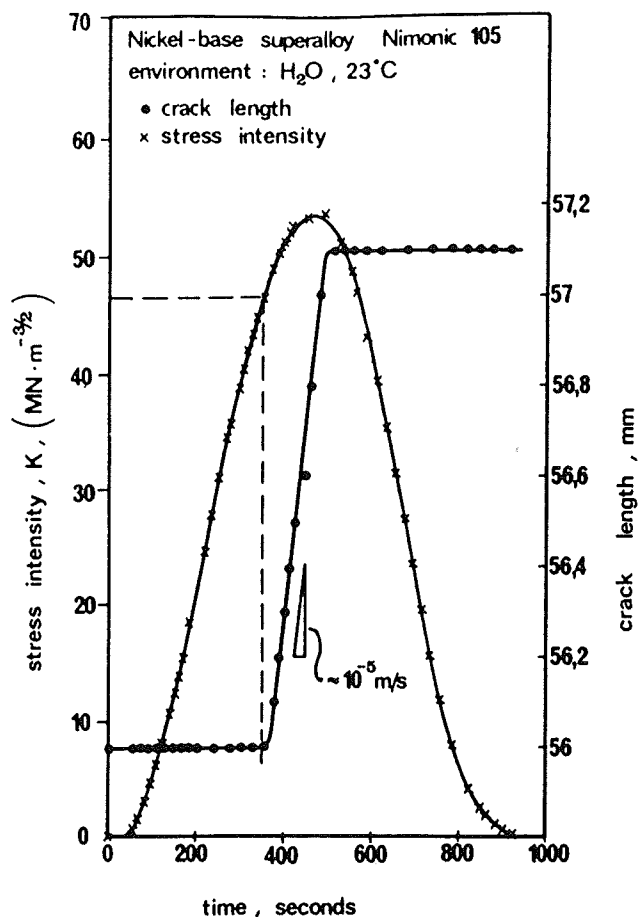
(c)

Stress-corrosion cracking behavior of nickel with phosphorus and sulfur segregation. (a) Polarization curve for nickel in 1*N* H₂SO₄ at 25 °C (77 °F). (b) Strain-to-failure and percent intergranular fracture for 26% phosphorus segregation at grain boundaries. (c) Strain-to-failure and percent intergranular fracture for 40% sulfur segregation at grain boundaries.



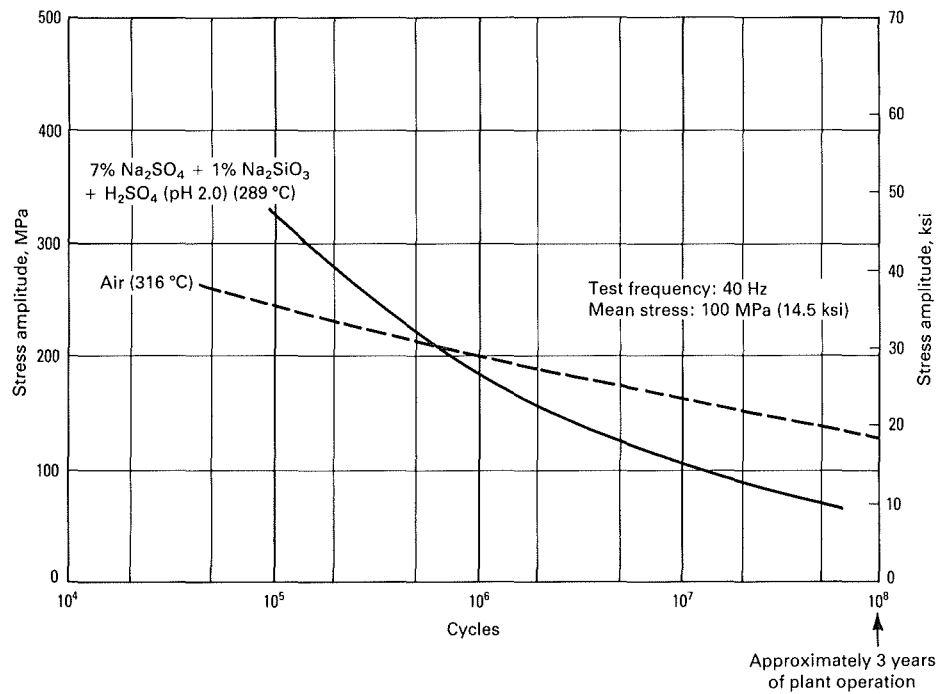
Effect of oxygen and water vapor pressure on fatigue life at 300 °C for 99.99% nickel at a plastic bending strain of 0.17%.

Nimonic 105: Corrosion Fatigue Crack Growth During Opening Part of a Load Cycle



Corrosion fatigue crack growing only during the opening part of a load cycle.

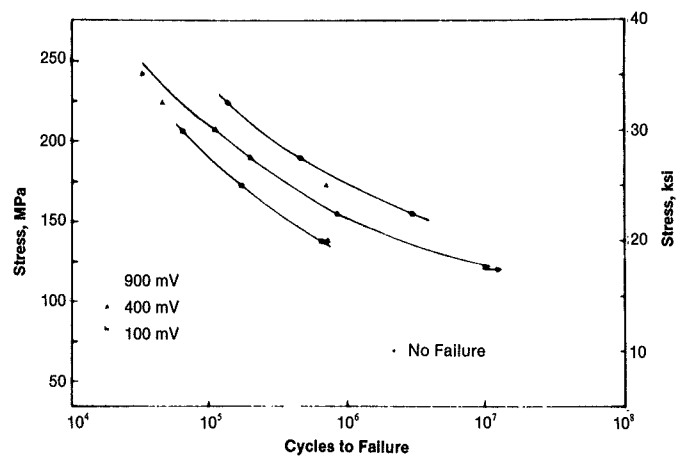
Composition: 15% Cr, 20% Co, 5% Mo, 4.5% Al, 1.4% Ti, 0.15% C, bal Ni. The figure shows that crack advance takes place only during the rising load portion of a tensile cycle.



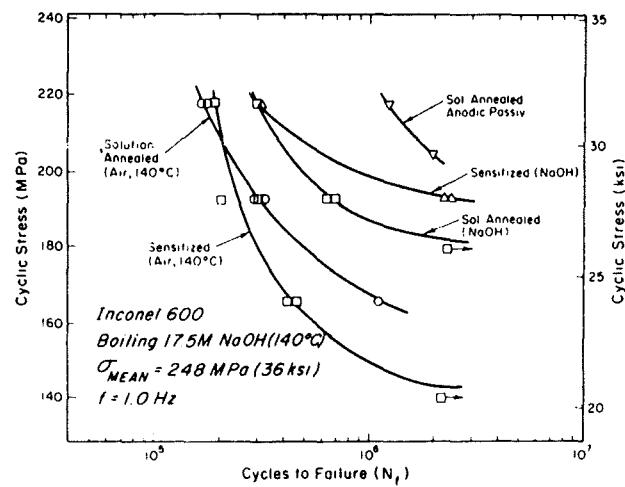
Corrosion fatigue behavior of Alloy 600.

In some PWRs, once-through steam generators are used, and cracking of some Alloy 600 tubes from the outer surface has occurred in upper regions. It is believed that these cracks are caused by corrosion fatigue resulting from small-amplitude vibration combined with impurities including acid sulfates/silicates. Corrosion fatigue tests show that in this environment the fatigue strength at long lifetimes is seriously degraded.

Nickel Alloy 600: S/N Curves, in NaCl Solution

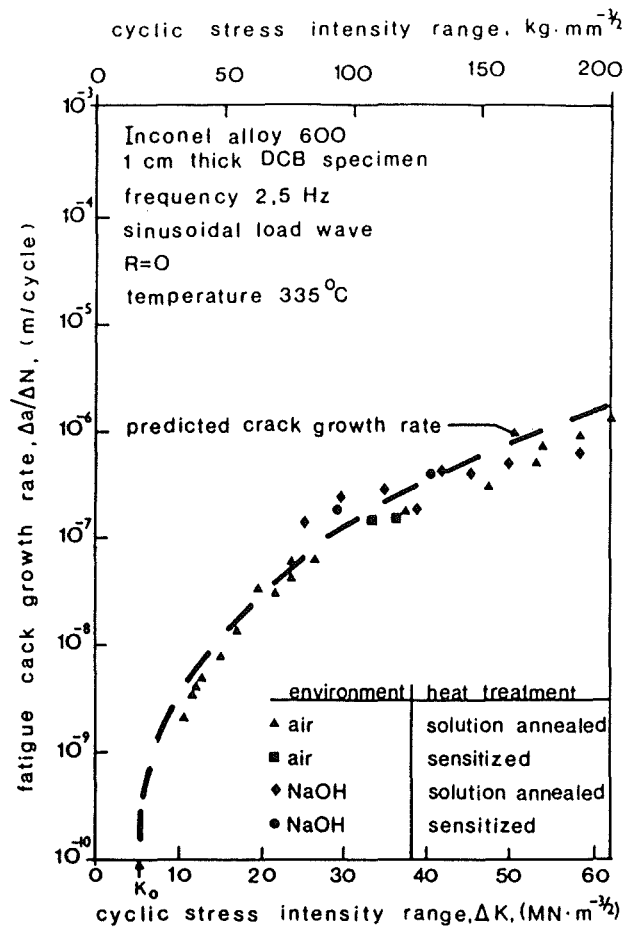


S/N curves for solution annealed Alloy 600 in deaerated 4M NaCl solution at different potentials.



S/N curves of Alloy 600 at 140°C in air and in boiling 17.5M NaOH solution.

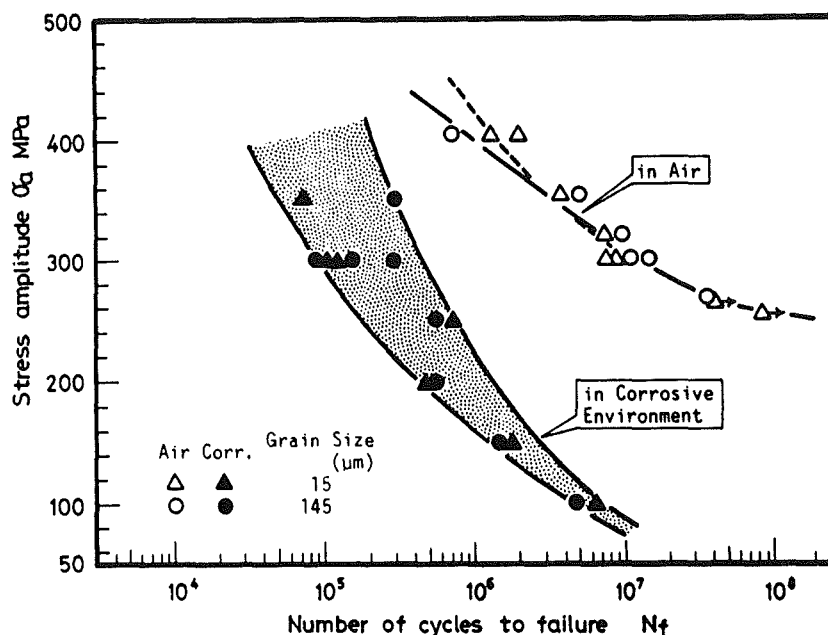
Nickel Alloy 600: Effect of Stress-Intensity Amplitude on Fatigue Crack Growth Rate, in Various Environments



Effect of stress-intensity amplitude and various environments on the fatigue crack growth rate of Inconel 600.

Composition: 76.0% Ni, 0.08% C, 0.5% Mn, 8.0% Fe, 0.008% S, 0.25% Cu, 25.5% Cr.

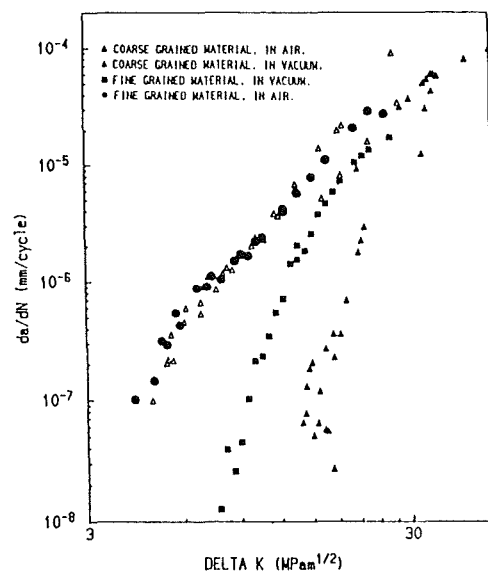
Source: Stress Corrosion Cracking and Hydrogen Embrittlement of Iron Base Alloys, NACE 5, National Association of Corrosion Engineers, Houston, 1974.



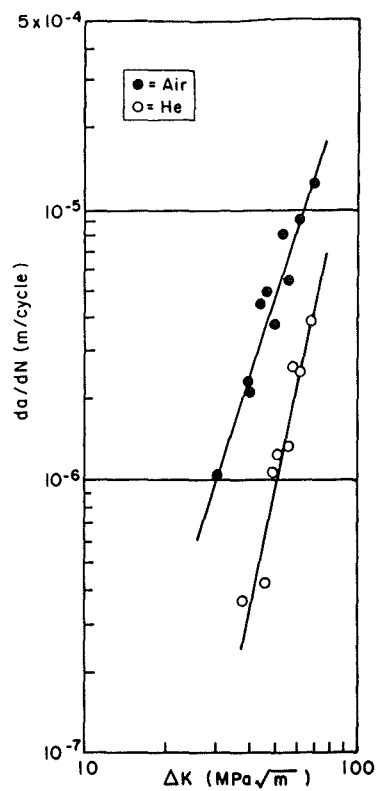
S/N curves of Inconel 718 specimens with different grain sizes at 800 °C in air and in a hot corrosive environment.

Composition: 0.03% C, 0.11% Si, 0.22% Mn, 0.011% P, 0.005% S, 17.54% Cr, 0.92% Ti, 0.42% Al, 3.02% Mo, 18.54% Fe, 4.99% Ta, + Nb, bal Ni. The fatigue tests were conducted at 800 °C using rotating bending fatigue machines operating at 1500 rpm. To simulate an actual hot corrosive environment, the specimens were coated with a synthetic salt mixture composed of 90% Na_2SO_4 plus 10% NaCl (melting point 785 °C). The amount of salty mixture applied was 0.4 kg/mm². In prolonged fatigue tests, the coating was reapplied at intervals of 3×10^5 cycles.

Inconel 718: Effect of Environment on Fatigue Crack Growth Rates



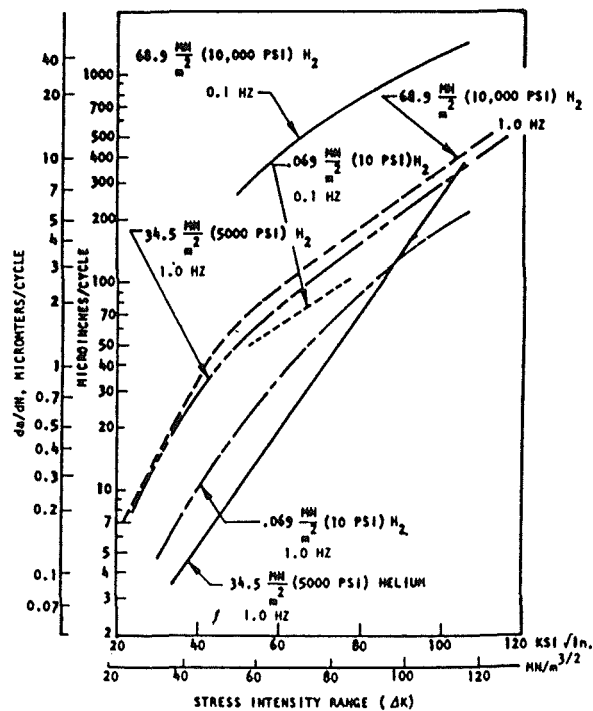
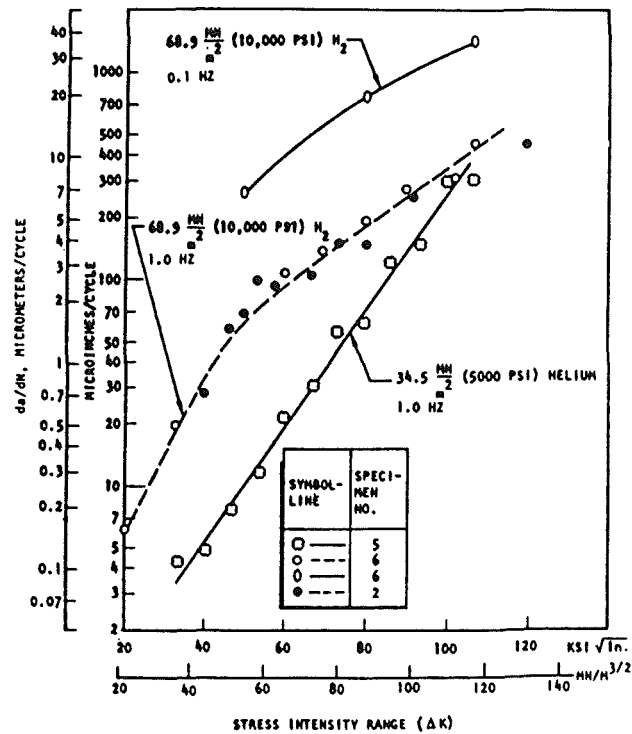
Effect of environment at 20 °C on fatigue crack growth rates in Inconel 718.



Fatigue crack growth data in air and in helium for Alloy 718 at 650 °C.

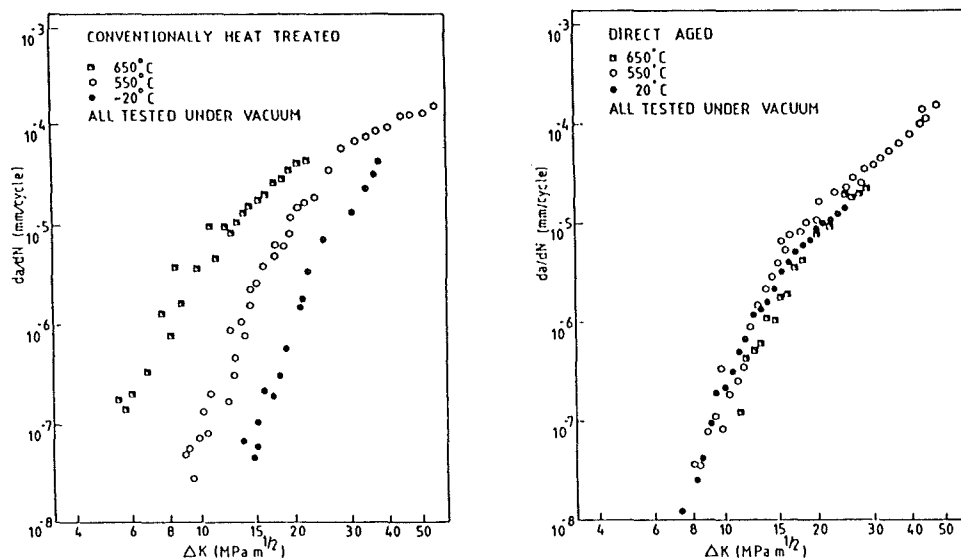
Results similar to those obtained in helium were obtained in N_2 , H_2 , + 2% CH_4 , and CO_2 atmospheres. The addition of 5% H_2O or 100 ppm O_2 to helium resulted in a crack growth rate similar to that obtained in air. The addition of 0.5% H_2S or 5% SO_2 to helium resulted in a fatigue crack growth rate higher than that found in air.

Inconel 718: Cyclic Crack Growth Rate, in Helium and Hydrogen



Cyclic crack growth rate for Inconel 718 in high-pressure helium and in high-pressure hydrogen at ambient temperature ($R = 0.1$).

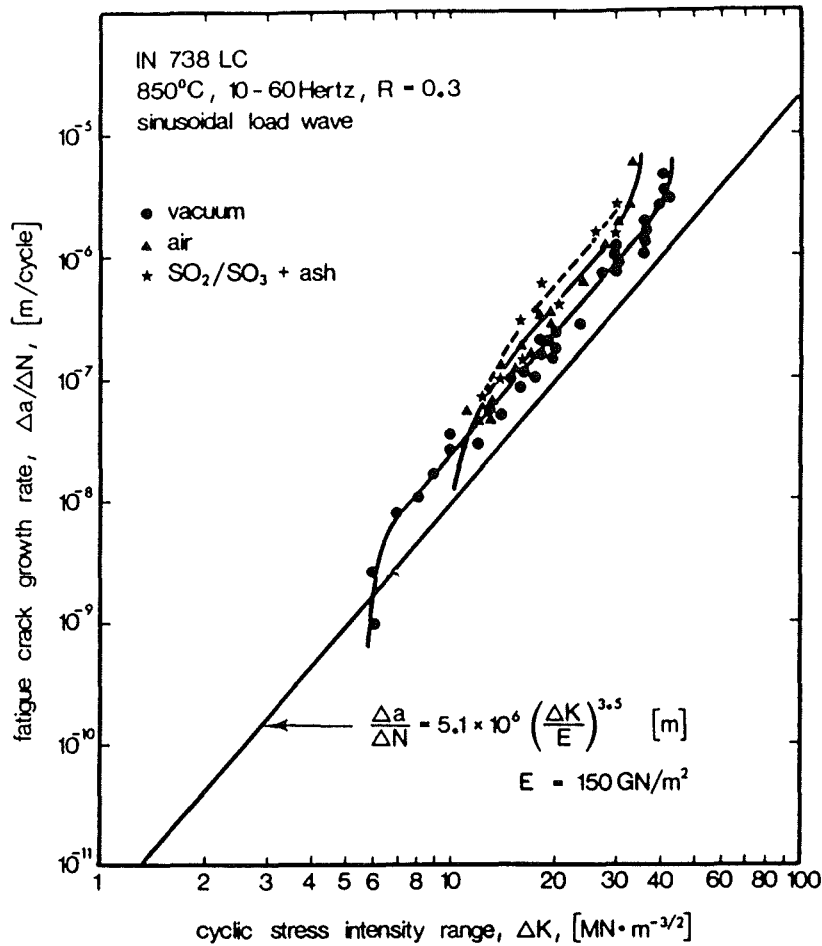
Source: R.J. Walter and W.T. Chandler, Environmental Degradation of Engineering Materials, M.R. Louthan, Jr. and R.P. McNitt, Ed., Virginia Tech., 1977, p 513-522.



Effect of temperature on the rate of fatigue crack growth in vacuum of Inconel 718. (a) Coarse grained material. (b) Fine grained material.

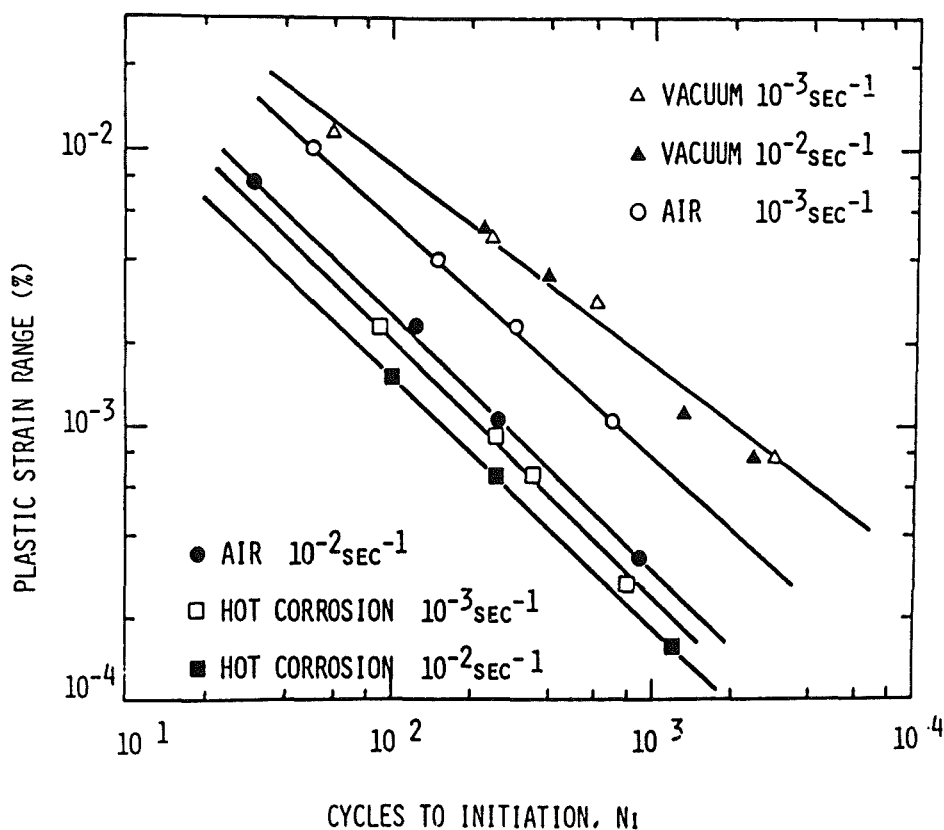
The specimens were subjected to cyclic loading at 5 Hz at 650 °C and 40 Hz at other temperatures. $R = 0.1$. The effect of temperature on the coarse grained material was attributed to differences in the roughness of the fracture surfaces, which affected closure levels.

Nickel Alloy IN 738 LC: Fatigue Crack Growth as a Function of Cyclic Stress-Intensity Range, in Various Environments



Fatigue crack growth in IN 738 LC at 850 °C as a function of the cyclic stress-intensity factor range in various environments.

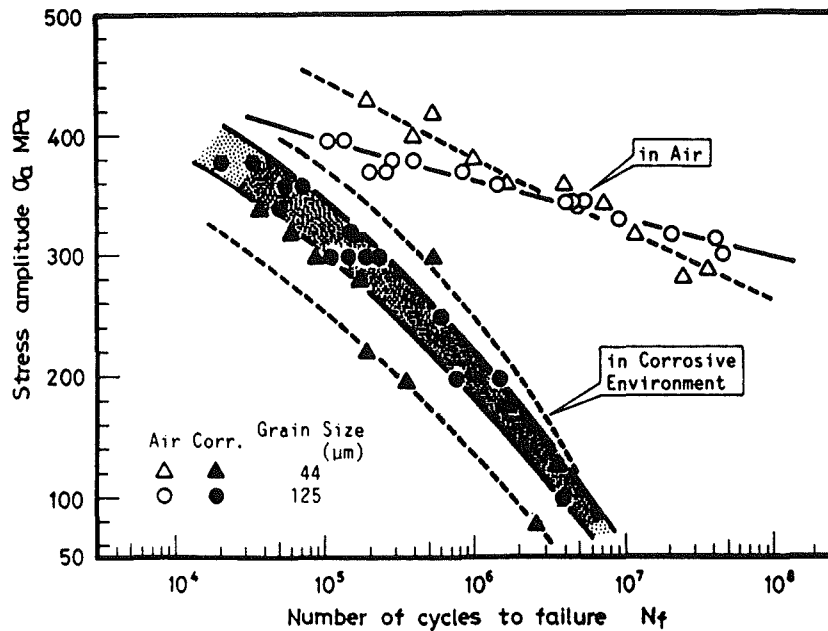
IN 738 LC: 0.17C-16.0Cr-8.5Co-1.7Mo-2.6W-1.7Ta-0.9Nb-3.4Al-3.4Ti-0.10Zr bal Ni. Heat treatment: 2 h at 1120 °C, AC + 24 h at 845 °C, AC. The simple power law line is shown for comparison. The SO₂/SO₃ + ash environment consisted of synthetic ash + air + 0.03 vol% SO₂/SO₃.



Effect of vacuum, air, and hot corrosive environments on the low-cycle fatigue behavior of IN738LC at 900 °C.

Composition: 8.38% Co, 15.9% Cr, 3.40% Al, 1.72% Ta, 2.50% W, 1.60% Mo, 3.30% Ti, 0.96% Nb, 0.07% Zr, 0.012% B, 0.12% C, bal Ni. For the hot corrosion tests, the specimens were coated with a mixture of 25% NaCl + 75% Na₂SO₄ by preheating to 150 °C and then rotating the specimens in a saltwater spray until an even coating of 9.5 mg/cm² was obtained. In the low-cycle fatigue tests, a triangular waveform was used with a mean strain of zero. Diametral strain was controlled with strain rates in the range of 10^{-2} to 10^{-3} per second.

Nickel Alloy 751: S/N Curves, in Air and in a Hot Corrosive Environment

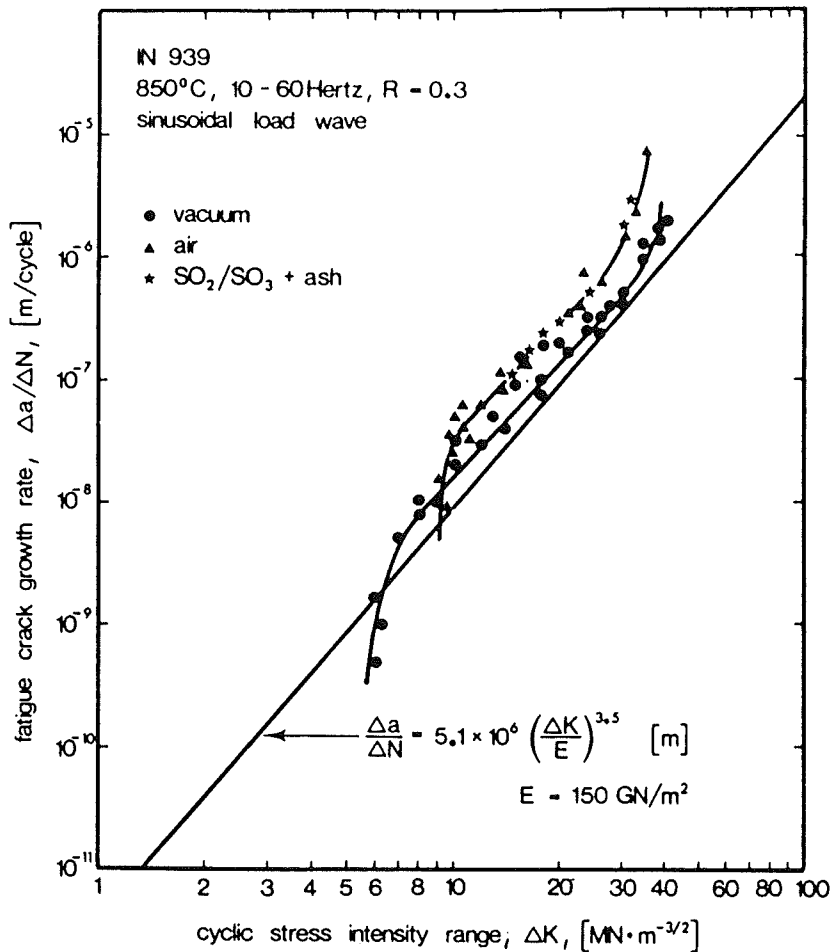


S/N curves of Inconel 751 specimens with different grain sizes at 800 °C in air and in a hot corrosive environment.

Composition: 0.09% C, 0.20% Si, 0.53% Mn, 0.007% S, 16.07% Cr, 2.05% Ti, 1.10% Al, 5.74% Fe, 1.18% Ta + Nb, bal Ni. The fatigue tests were conducted at 800 °C using rotating bending fatigue machines operating at 1500 rpm. To simulate an actual hot corrosive environment, the specimens were coated with a synthetic salt mixture composed of 90% Na_2SO_4 plus 10% NaCl (melting point 785 °C). The amount of salty mixture applied was 0.4 kg/mm². In prolonged fatigue tests, the coating was re-applied at intervals of 3×10^5 cycles.

Nickel Alloy IN 939: Fatigue Crack Growth Rate as a Function of Cyclic Stress-Intensity Range, in Various Environments

431



Fatigue crack growth rate in IN 939 at 850 °C as a function of the cyclic stress-intensity factor range in various environments.

IN 939: 0.15C-22.6Cr-19.1Co-0.05Mo-2.0W-1.0Ta-1.0Nb-1.9Al-3.7Ti-0.1Zr bal Ni. Heat treatment: 4 h at 1160 °C, fast air cool (FAC) + 24 h at 900 °C, AC + 16 h at 700 °C, AC.

TITANIUM ALLOYS

TITANIUM ALLOYS

The designation and nominal compositions of a number of titanium alloys is given in the table below. The unalloyed grades are immune to stress-corrosion cracking (SCC) except for a few specific environments such as anhydrous/methanol/halide solutions, nitrogen tetroxide, red fuming nitric acid, and liquid or solid cadmium. Other titanium alloys—including the aerospace alloys—may also be susceptible in aqueous chloride solutions as well as in additional environments. However, this susceptibility is almost always associated with the high stress concentrations typical of the precracked specimens used in laboratory testing, and is generally not observed with smooth or mildly notched specimens. Fortunately, the susceptibility identified for these alloys in laboratory tests is seldom observed in actual applications, although the potential for SCC is there should fatigue cracks develop in service. Both anodic dissolution and anodic dissolution accompanied by hydrogen assisted cracking have been proposed as mechanisms of SCC in titanium alloys in aqueous environments. Hydride formation may also be involved in delayed fracture mechanisms.

Commercial Titanium and Titanium Alloys: Environments Known to Promote Stress-Corrosion Cracking

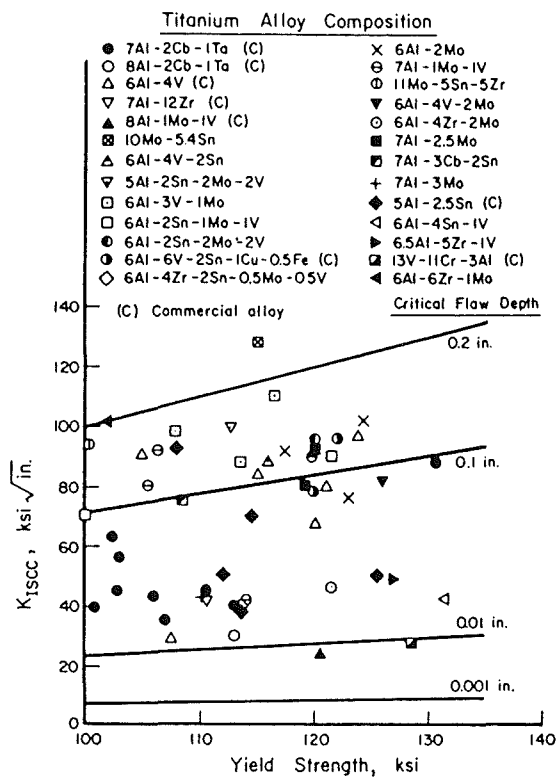
437

Environments known to promote SCC of commercial titanium and titanium alloys

Medium	Temperature		Susceptible materials
	°C	°F	
Oxidizers			
Nitric acid, RFNA.....		Room	Unalloyed titanium, Ti-8Mn, Ti-6Al-4V, Ti-5Al-2.5Sn, Ti-2Fe-2Cr-2Mo
Nitrogen tetroxide (no excess NO).....	30-75	85-165	Ti-6Al-4V
Organic compounds			
Methanol (chloride, bromide)		Room	Ti-6Al-4V Ti-6Al-4V, Ti-8Al-1Mo-1V, grade 4 Ti-6Al-4V, Ti-8Al-1Mo-1V, Ti-5Al-2.5Sn, Ti-4Al-3Mo-1V
Methyl chloroform, inhibited.....	370	700	Ti-8Al-1Mo-1V, Ti-6Al-4V, Ti-5Al-2.5Sn, Ti-13V-11Cr-3Al
Ethyl alcohol.....		Room	Ti-8Al-1Mo-1V Ti-8Al-1Mo-1V, Ti-5Al-2.5Sn
Ethylene glycol.....		Room	Ti-8Al-1Mo-1V
Trichloroethylene.....	370, 620, 815	700, 1150, 1500	Ti-8Al-1Mo-1V, Ti-5Al-2.5Sn Ti-5Al-2.5Sn
Trichlorofluoroethane	788	1450	Ti-8Al-1Mo-1V, Ti-5Al-2.5Sn, Ti-6Al-4V, Ti-13V-11Cr-3Al
Chlorinated diphenyl.....	315-370	600-700	Ti-5Al-2.5Sn
Hot salt			
Chloride salts—various, residues	288-426	550-800	All commercial alloys except grades 1, 2, 7, 11, and 12
Metal embrittlement			
Cadmium		Room	Ti-6Al-4V
	329-400	625-750	Ti-8Mn
Mercury.....		Room	Grade 4 Ti, Ti-6Al-4V
	370	700	Ti-13V-11Cr-3Al
Silver			
Plate	470	875	Ti-7Al-4Mo, Ti-5Al-2.5Sn
AgCl	370-480	700-900	Ti-7Al-4Mo, Ti-5Al-2.5Sn
Ag-5Al-2.5Mn.....	340	650	Ti-6Al-4V, Ti-8Al-1Mo-1V
Seawater		Ambient	Unalloyed titanium (with high oxygen content) Ti-8Mn Ti-2.5Al-1Mo-11Sn-5Zr-0.2Sc Ti-3Al-11Cr-13V Ti-5Al-2.5Sn Ti-6Al-4V Ti-6Al-6V-2Sn Ti-8Al-1Mo-1V Ti-3Al-8V-6Cr-4Zr-4Mo
Miscellaneous			
Chlorine.....	288	550	Ti-8Al-1Mo-1V
Hydrochloric acid			
10%.....	35	95	Ti-5Al-2.5Sn
.....	340	650	Ti-8Al-1Mo-1V

Environments known to promote stress-corrosion cracking of commercial titanium and titanium alloys.

Titanium Alloys: Stress-Corrosion Cracking Resistance Index, in Salt Water

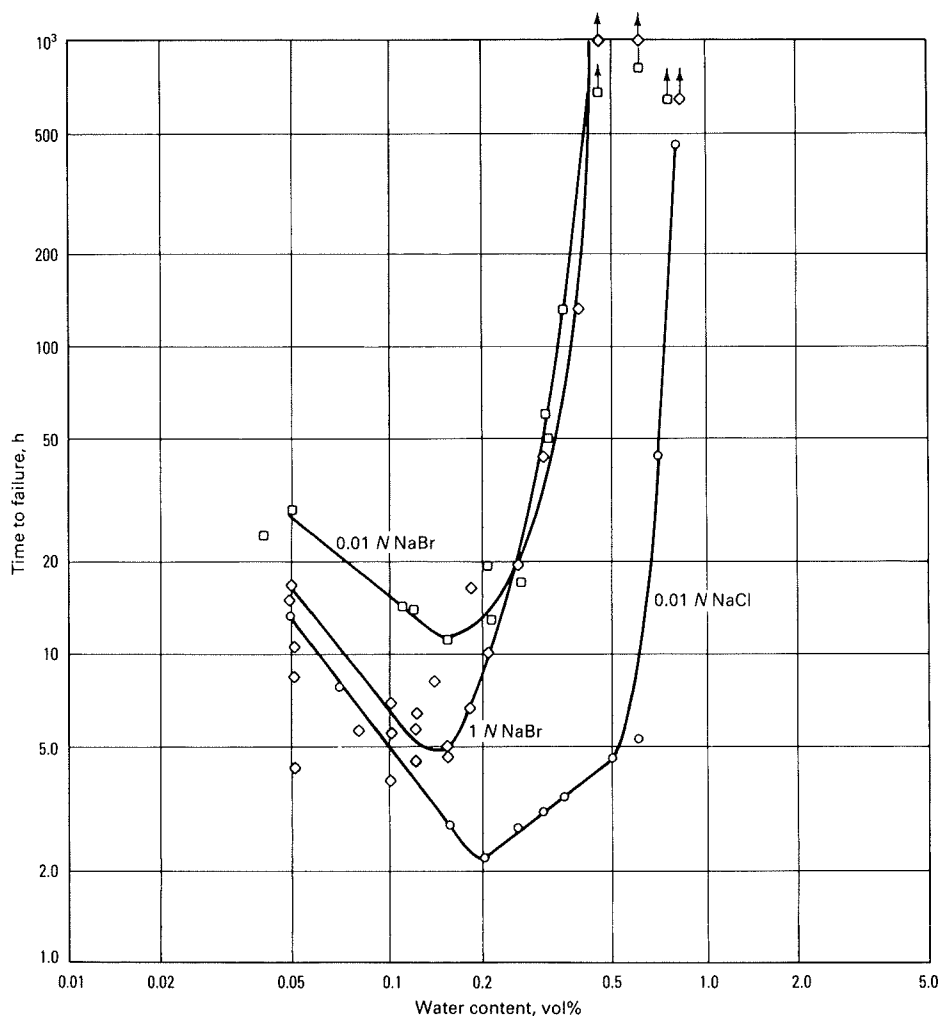


Stress-corrosion cracking resistance index for titanium alloys in 3.5% salt water.

The critical flaw depth is obtained from the relation $0.22\sigma_y a^{1/2} = K_{Isc}$. This expression assumes that the applied stress is at the yield stress level and that a surface crack of length $2c$ and of depth a is present, where a is much less than c . For a given yield stress and flaw size, the value of K_{Isc} must exceed that given by this relation if failure is to be avoided.

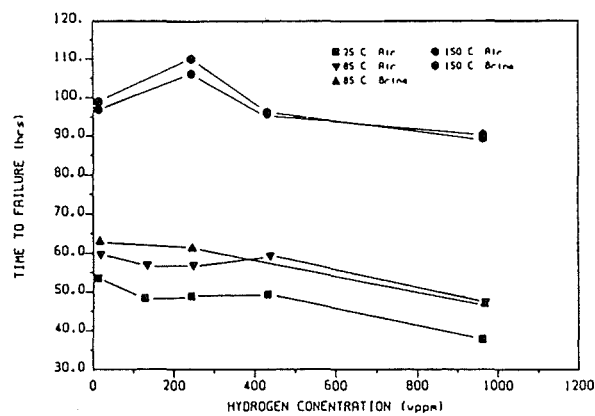
Commercially Pure Titanium: Effect of Bromide and Chloride Additions on Stress-Corrosion Cracking, in Methanol/Water Solutions

439



Effect of bromide and chloride additions on stress-corrosion cracking of cold rolled and annealed commercially pure titanium stressed to 75% of yield strength in methanol/water solutions.

Titanium Grade 12: Effect of Hydrogen Concentration on Time-to-Failure, in Air and in Brine



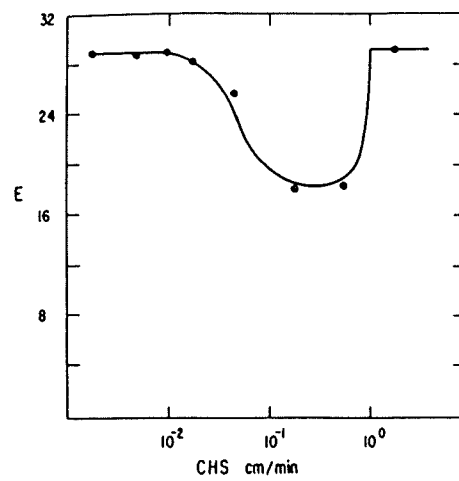
Effect of hydrogen concentration on time-to-failure of smooth tensile specimens of Titanium Grade 12. The tests were run in air and in brine over a temperature range of 25 to 150 °C.

Composition, wt%						
Ni	Mo	Fe	O	N	Ti	H (ppm)
0.697	0.215	0.050	0.075	0.010	bal	27

Source: N.R. Sorensen and J.A. Ruppen, The Environmental Cracking of Ticode-12 in Aqueous Chloride Environments, Corrosion, 41(10), October 1985, p 560-569.

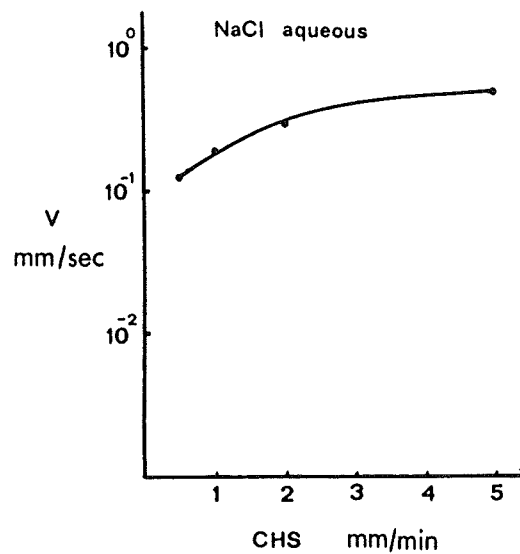
Ti-O Alloy: Elongation-to-Fracture as a Function of Crosshead Speed, in $\text{CH}_3\text{OH} + \text{HCl}$

441

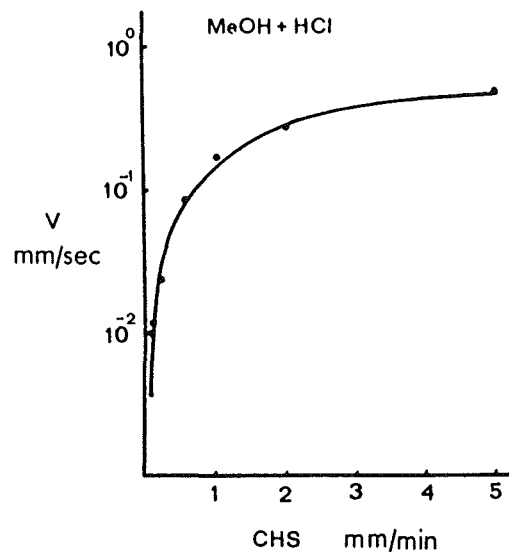


Elongation-to-fracture of a tensile specimen of a Ti-O alloy with a gauge length of 3 cm in a $\text{CH}_3\text{OH} + \text{HCl}$ mixture as a function of crosshead speed.

Ti-O Alloy: Stress-Corrosion Cracking Velocity as a Function of Crosshead Speed, in Aqueous Sodium Chloride Solution



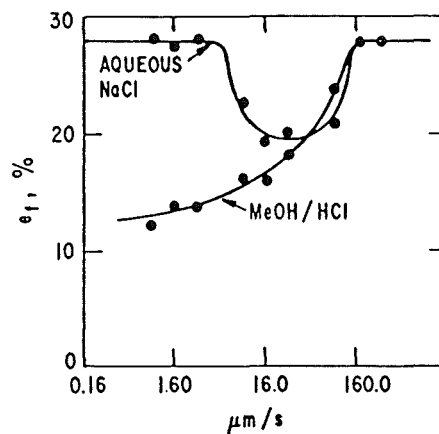
Stress-corrosion cracking velocity in a Ti-O alloy in a 3% aqueous sodium chloride solution as a function of crosshead speed.



Stress-corrosion cracking velocity in a Ti-O alloy tested in a CH₃OH + HCl mixture as a function of crosshead speed.

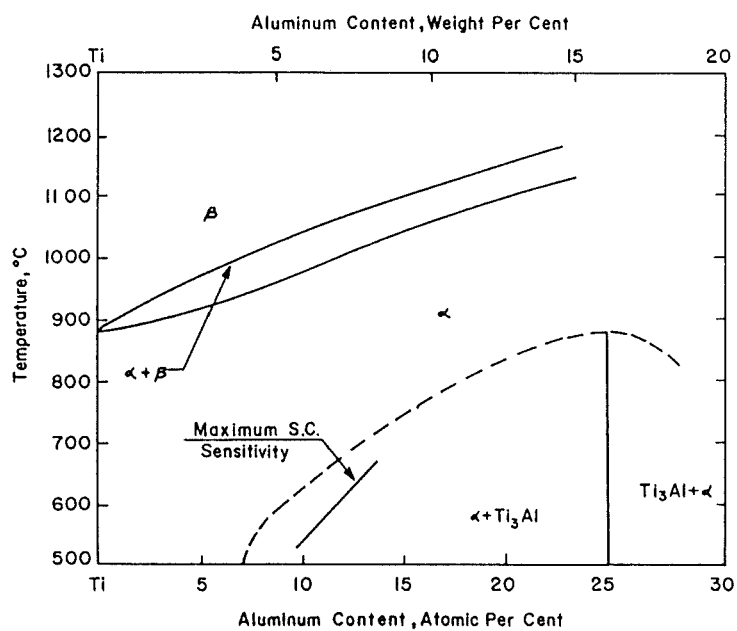
Note that the crack velocity depends on the crosshead speed. If the crosshead speed were to be lowered, the crack velocity would also decrease to a rate corresponding to the new crosshead speed. However, below a minimum crosshead speed, the crack will arrest, which is an indication that the effective strain rate at the tip of a propagating crack must be above a certain value to prevent repassivation. K_{ISCC} would appear to be that value of K which produces a crack tip strain rate less than the critical value in a particular environment.

Ti-O Alloy: Effect of Extension Rate on Elongation-to-Fracture, in Aqueous Sodium Chloride and in Acidic Methanol (HCl)



Effect of extension rate on elongation-to-fracture of a Ti-O alloy in aqueous sodium chloride and in acidic methanol (HCl).

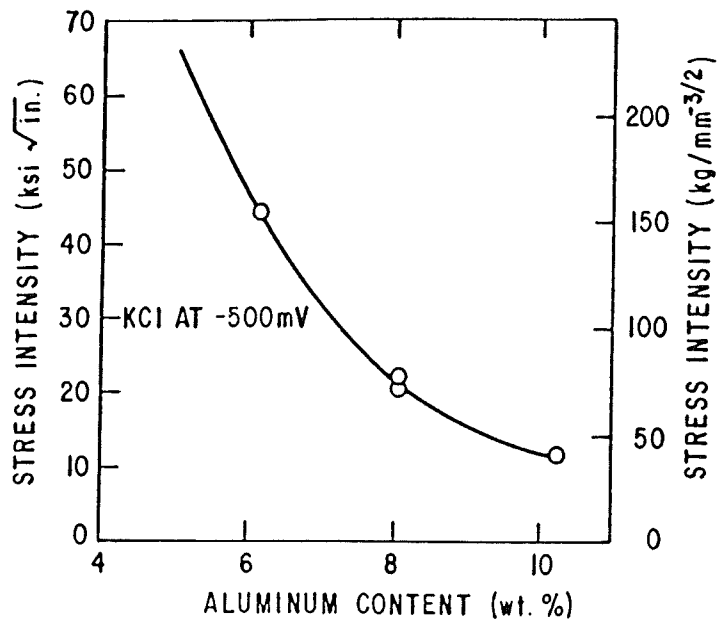
The corrosion of titanium and titanium alloys in methanol/halogen solutions differs from that in sodium chloride solutions in a number of ways. Titanium of various purity levels, as well as a number of α alloys, will corrode intergranularly if left unstressed in a methanol/halogen solution. If the halogen is not present, corrosion in unstressed samples will not occur. When samples are strained at very low strain rates in methanol/halogen environments, a severe degradation is observed, whereas in aqueous sodium chloride passivation occurs rapidly enough to prevent degradation of mechanical properties. Alloys containing the β phase usually do not corrode in the unstressed condition, but do so readily when stressed. The susceptibility to stress-corrosion cracking increases with a decrease in elongation at fracture, e_f , in constant extension rate tests.



Titanium-aluminum phase diagram.

The aluminum threshold for sensitivity agrees fairly well with the formation of Ti_3Al .

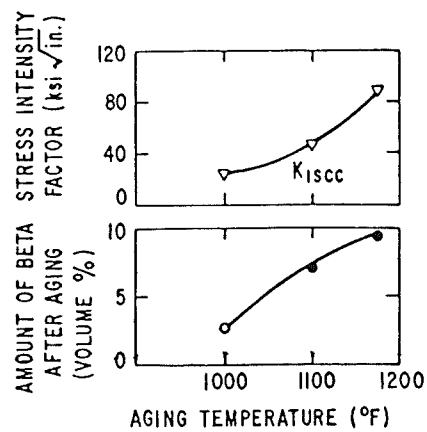
Ti-Al Alloys: Effect of Aluminum on Threshold Stress-Intensity Factor, in Aqueous KCl



Effect of aluminum on the threshold stress-intensity factor, K_{Isc} , for stress-corrosion cracking of Ti-Al alloys in aqueous KCl.

Ti-4Al-3Mo-1V Titanium Alloy: Effect of Increasing Volume Percent of Beta Phase on K_{Isc}

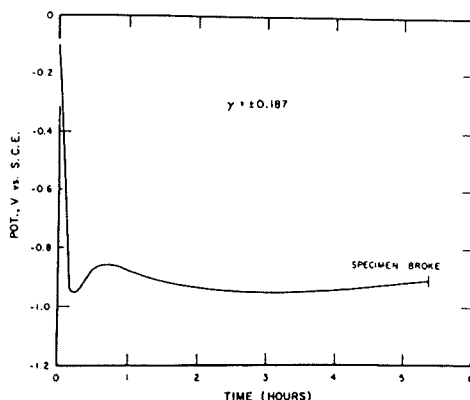
447



Effect of increasing the volume percent of β phase on K_{Isc} in Ti-4Al-3Mo-1V.

Environment: 3.5% sodium chloride solution, pH 6.5.

Ti-4.5Al-5Mo-1.5Cr Titanium Alloy: Open Circuit Potential Versus Time Curve, in Hanks Solution



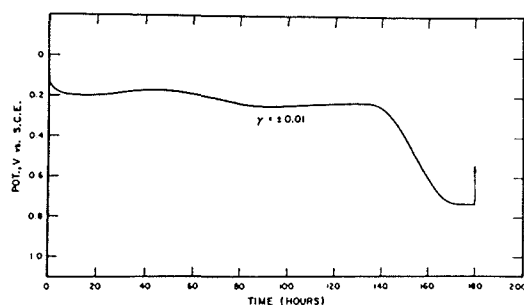
Open circuit potential versus time curve for Corona 5 (Ti-4.5Al-5Mo-1.5Cr) during corrosion fatigue testing at a shear strain range of ± 0.0187 in Hanks solution (pH 7.4) at 37 °C.

Corona 5 was developed to obtain a higher fracture toughness than Ti-6Al-4V. It is richer in β -stabilizers than Ti-6Al-4V and has a β -transus about 55 °C lower.

The measurement of the open circuit electrode potential during cyclic loading provides information on the formation or breakdown of the surface oxide film. If the electrode potential rises with cycling, the indication is that a surface film is forming. If the potential falls, the surface film is being broken down by dissolution, or is being cracked by mechanical processes. A rise after a fall indicates that repassivation is occurring. Frequency is approximately 1 Hz. Number of cycles to failure is 1.8×10^4 . Composition of Hanks solution: 8.00 g/L NaCl, 0.14 g/L CaCl_2 , 0.40 g/L KCl, 0.35 g/L NaHCO_3 , 1.00 g/L glucose, 0.10 g/L $\text{MgCl}_2 \cdot 6\text{H}_2\text{O}$, 0.06 g/L $\text{Na}_2\text{HPO}_4 \cdot 2\text{H}_2\text{O}$, 0.06 g/L $\text{MgSO}_4 \cdot 7\text{H}_2\text{O}$, 0.10 g/L NaH_2PO_4 .

Ti-4.5Al-5Mo-1.5Cr Titanium Alloy: Open Circuit Potential Versus Time Curve, in Hanks Solution

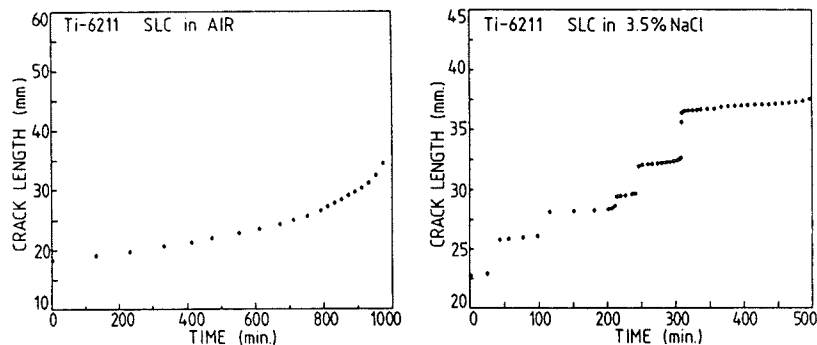
449



Open circuit potential versus time curve for Corona 5 (Ti-4.5Al-5Mo-1.5Cr) during corrosion fatigue testing at a shear strain range of ± 0.01 in Hanks solution (pH 7.4) at 37 °C.

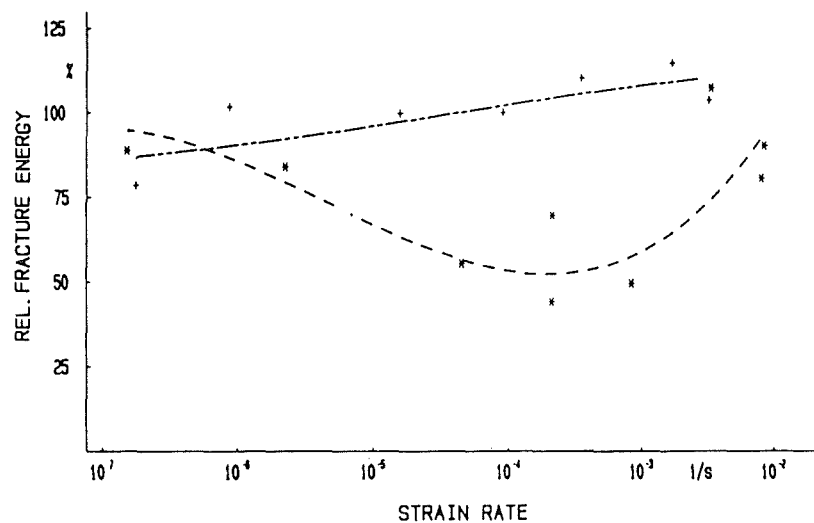
Frequency is approximately 1 Hz. Number of cycles to failure is 6.5×10^5 .

Ti-6Al-2Cb-1Ta-0.8Mo Titanium Alloy: Crack Length Versus Time, in Air and in Sodium Chloride



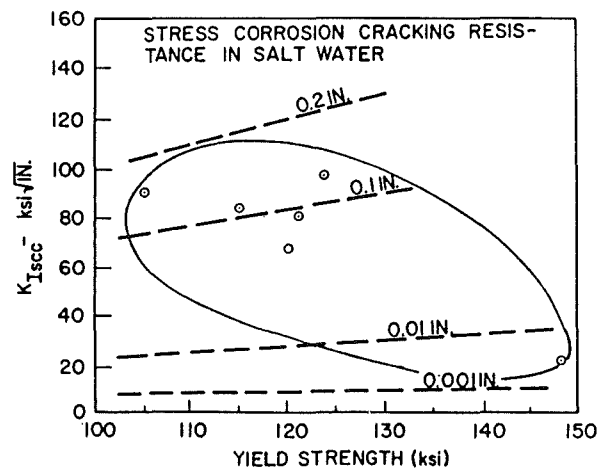
Plots of crack length versus time that illustrate continuous sustained load crack growth in air and discontinuous sustained load crack growth in 3.5% sodium chloride solution.

It has been proposed that sustained load cracking of titanium alloys in aqueous salt water environments results from the diffusion of hydrogen from the environment into the crack tip region. The increased hydrogen concentration in the hydrostatic stress field ahead of the crack tip results in hydride precipitation and subsequent cleavage fracture associated with the hydrides. Discontinuous crack growth occurs each time the crack moves out of the hydride-affected region and moves into hydride-free material where it is arrested.



Stress-corrosion cracking behavior of smooth specimens of P/M Ti-6Al-4V. Dashed line (stars) represent powder sprayed under vacuum (PSV-powder). Long dash-dashes line (crosses) represent plasma rotating electrode process (PREP-powder). The experiments were performed in 3.5% sodium chloride solution at 25 °C.

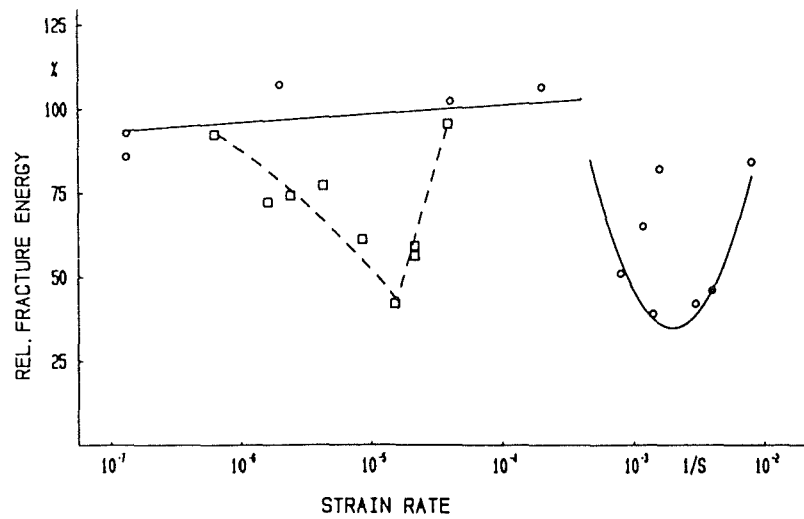
Ti-6Al-4V Titanium Alloy: K_{Isc} Characteristics at Various Strength Levels, in Salt Water



K_{Isc} characteristics of various lots of Ti-6Al-4V at various strength levels in salt water. The oval encloses the known K_{Isc} data for this alloy in salt water.

Ti-6Al-4V Titanium Alloy: Stress-Corrosion Cracking Behavior, in Sodium Chloride

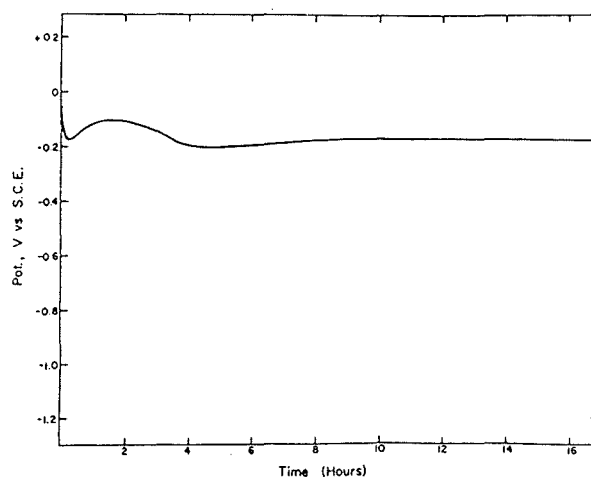
453



Stress-corrosion cracking behavior of I/M Ti-6Al-4V in the smooth (solid line) and precracked (dashed line) conditions in 3.5% sodium chloride.

The strain rates represent overall deformation rates and do not reflect the much higher strain rates at the crack front.

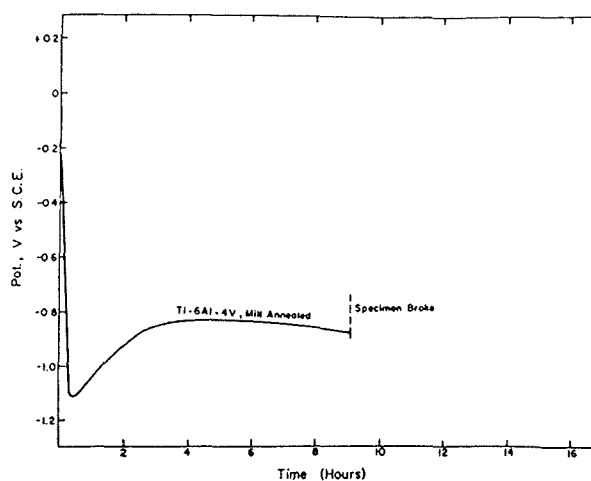
Ti-6Al-4V Titanium Alloy: Open Circuit Potential Versus Time Curve, in Hanks Solution



Open circuit potential versus time curve for Ti-6Al-4V during corrosion fatigue testing at a shear strain range of ± 0.01 in Hanks solution (pH 7.4) at 37 °C. Number of cycles to failure is 2.2×10^5 .

Ti-6Al-4V Titanium Alloy: Open Circuit Potential Versus Time Curve, in Hanks Solution

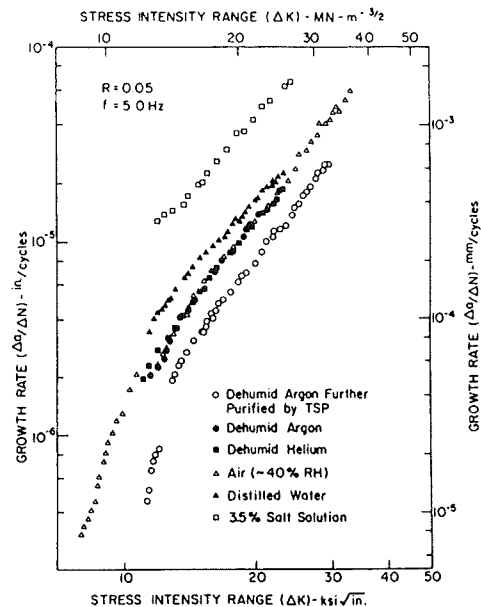
455



Open circuit potential versus time curve for Ti-6Al-4V as indicated during corrosion fatigue testing at a shear strain range of ± 0.0187 in Hanks solution (pH 7.4) at 37 °C. Number of cycles to failure is 3.5×10^4 .

Source: M.A. Imam, et al., Corrosion and Corrosion-Fatigue Behavior of Ti-4.5Al-5Mo-1.5Cr (Corona 5) and Ti-6Al-4V, in Titanium '80: Science and Technology, Vol 4, H. Kimura and O. Izumi, Ed., American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, PA, 1980, p 2595-2604.

Ti-6Al-4V Titanium Alloy: Kinetics of Fatigue Crack Growth, in Various Environments

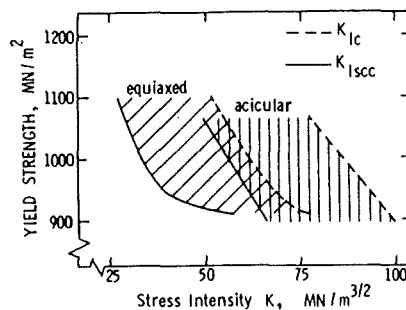


Kinetics of fatigue crack growth in mill annealed Ti-6Al-4V in various environments at room temperature.

The residual moisture levels in the dehumidified environments were well below 30 ppm. Dehumidification was accomplished by passing ultra-high-purity grade argon (99.999% purity) through a molecular sieve drier, cold traps at less than -140°C , and a titanium sublimation pump (TSP), which served as a getter and as a pump for active residual impurities in the test chamber.

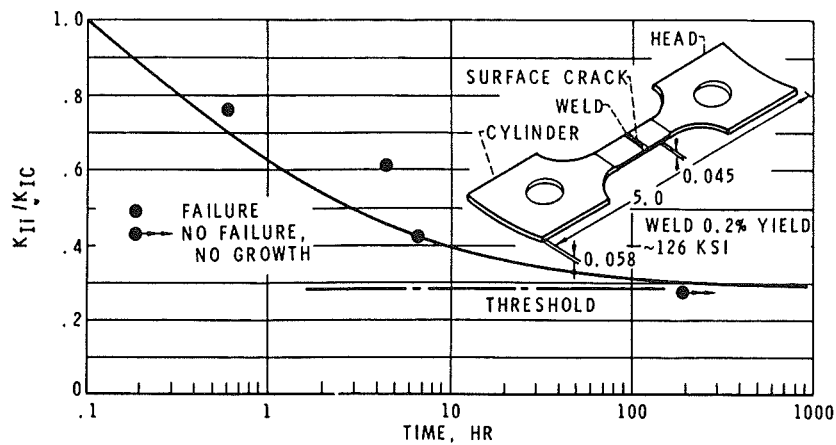
Ti-6Al-4V Titanium Alloy: Relationship Between Yield Stress and Stress-Intensity, in Salt Water

457



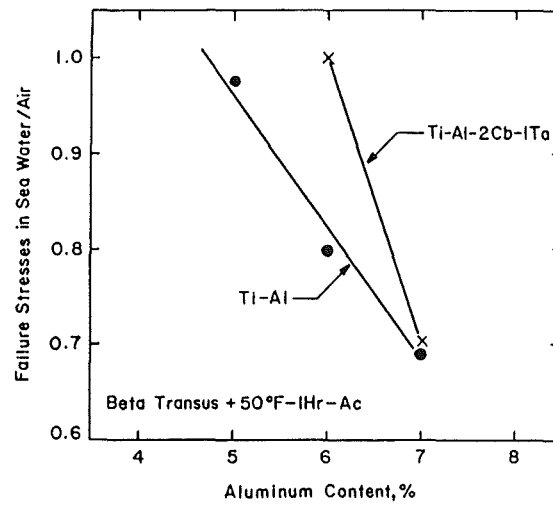
Relationship between yield stress and stress-intensity for either unstable fast fracture or stress-corrosion cracking in salt water for Ti-6Al-4V. A variety of microstructures for each type were tested.

Ti-6Al-4V Titanium Alloy: Stress-Corrosion Cracking, in Methanol



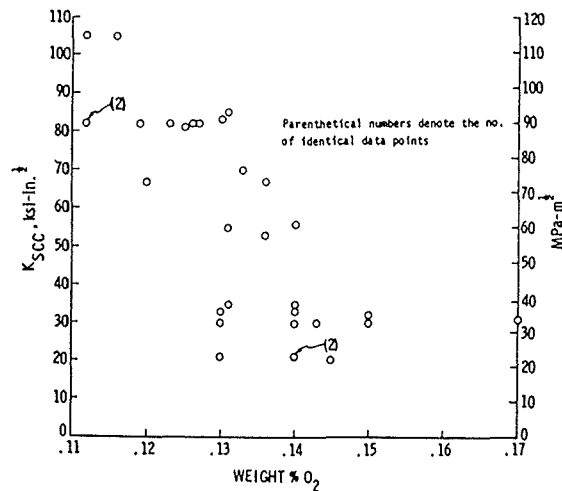
Stress-corrosion cracking results at room temperature in methanol for Ti-6Al-4V weldments.

Cracks were introduced into the heat-affected zones to localize the failures. Similar results were obtained in Freon.



Effect of aluminum content on aqueous stress-corrosion of Ti-6Al-4V plate.

Ti-6Al-4V Titanium Alloy: Transverse K_{SCC} as a Function of Oxygen Content



Transverse K_{SCC} as a function of oxygen content.

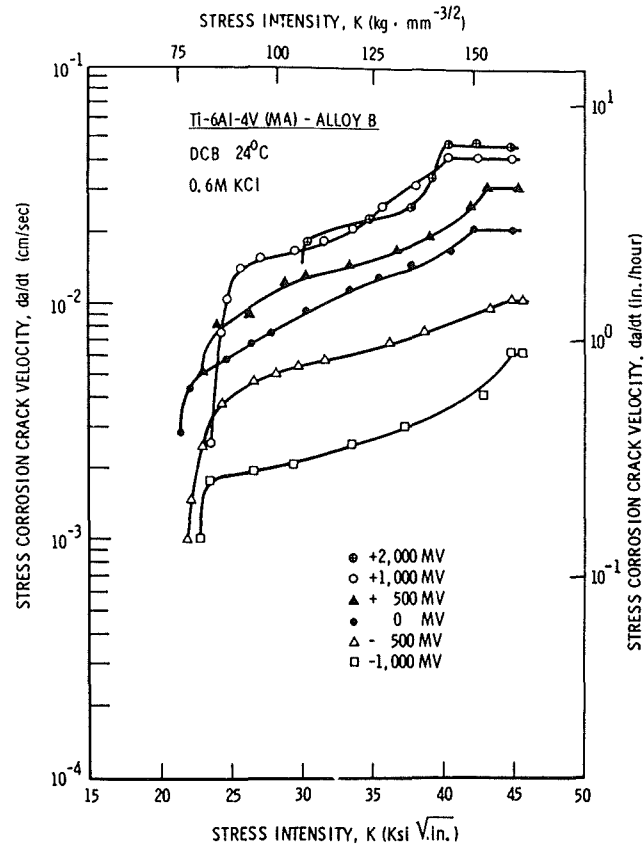
The data obtained in this study involved Ti-6Al-4V sheet from 46 commercial heats of duplex annealed material. Regression analyses were used to correlate K_{SCC} values to metallurgical parameters. The equation that exhibited the best fit was:

$$\text{Transverse } K_{SCC} \text{ ksi-in.}^{1/2} = 321.9 - 714 (O_2) - 22.8(Al) - 2.74(\Delta TYS) - 2.95 (O.F.) - 0.224(H_2)$$

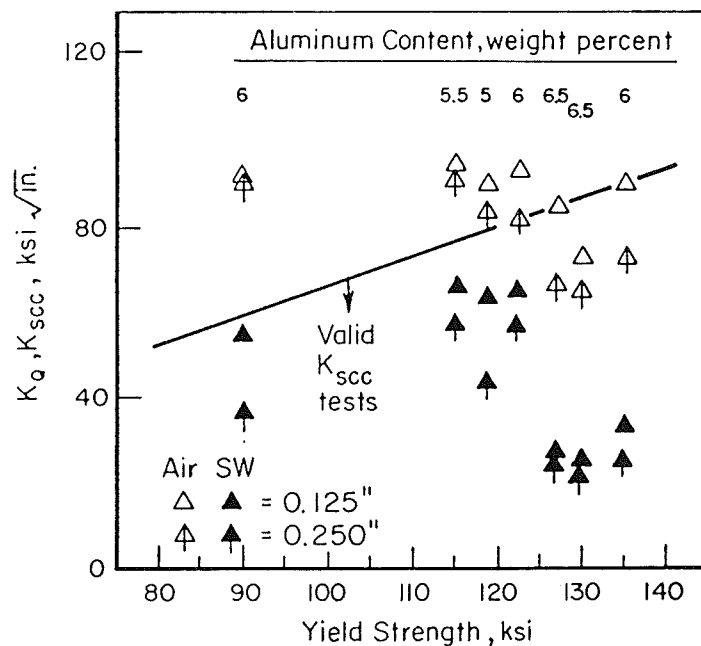
where O_2 is wt% oxygen, Al is wt% aluminum, ΔTYS is the transverse minus the longitudinal yield strength in ksi, a reflection of texture, O.F. is an ordering factor reflecting the degree of α -phase ordering that occurs on slow cooling through the $\alpha + \alpha_2$ (Ti_3Al) phase field, and hydrogen is the ppm of hydrogen by weight.

Ti-6Al-4V Titanium Alloy: Effect of Stress-Intensity and Potential on Stress-Corrosion Crack Velocity, in KCl Solution

461



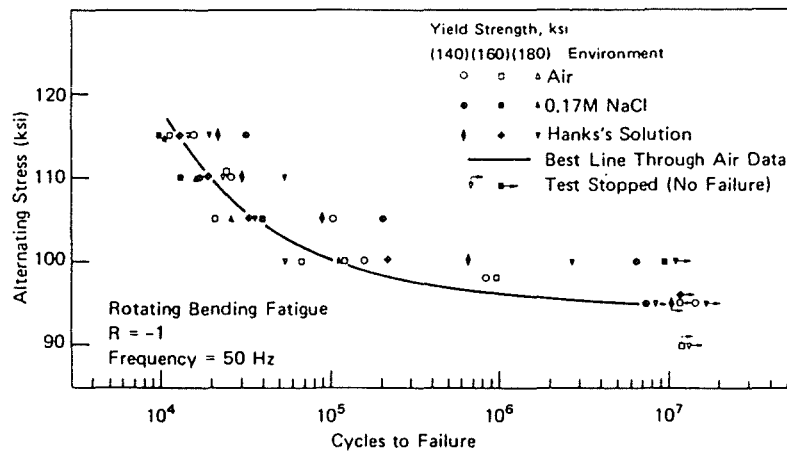
Effect of stress-intensity and potential on stress-corrosion crack velocity in Ti-6Al-4V tested in 0.6M KCl solution. Note that K_{Isc} exhibits a minimum value between 0 and -500 mV. The Region II stress-corrosion crack velocity increases in a similar manner to iodide solutions. Note, however, that the +2000 mV and +1000 mV curves are rather similar.

Effect of Aluminum Content on K_Q and K_{SCC} Effect of aluminum content on K_Q and K_{SCC} for 0.250-in. thick specimens.

The "validity line" for the salt water tests is given by the thickness, $B \geq 0.57(K/Y)^2$, where Y is the yield strength. According to ASTM E399, the requirement for plane-strain conditions to exist is that $B > 2.5 (K/\sigma_y)^2$. It has been argued that this requirement may be too conservative for measurements of K_{SCC} , because subcritical crack growth could occur when only a small fraction of the crack front is subjected to plane-strain loading. Piper and Feeney (ARPA Coupling Program on Stress-Corrosion Cracking, Eleventh Quarterly Report, July 1969, p 6) measured the stress-intensity required to initiate subcritical cracking in Ti-6Al-4V as a function of thickness and found that K_{SCC} remains unchanged for thicknesses in the range 0.10 to 1.0 in., i.e., $B > 0.57 (K_{SCC}/\sigma_y)^2$. An examination of the results shown in the above figure indicates that both K_Q and K_{SCC} were lower for the 0.250-in. thick specimens than for the 0.125-in. thick specimens. However, a distinct effect of aluminum on the susceptibility to stress-corrosion cracking was still apparent. It was concluded that although measurements of K_Q and K_{SCC} may not be quantitatively accurate due to mixed mode loading effects, it is valid to determine the stress-corrosion cracking susceptibility from a qualitative comparison of K_Q and K_{SCC} .

Ti-6Al-6V-2Sn Titanium Alloy: Effect of Environment on Fatigue Life

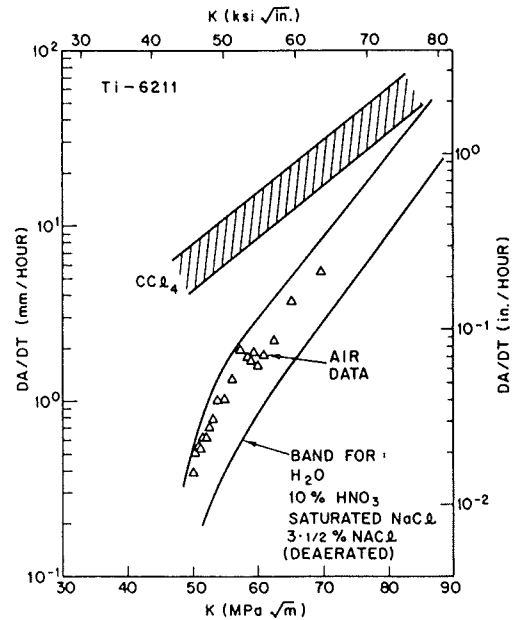
463



Effect of environment on the fatigue life of Ti-6Al-6V-2Sn.

Source: M. Levy, et al., The Corrosion Behavior of Titanium Alloys in Chloride Solutions: Materials for Surgical Implants, in Titanium Science and Technology, R.I. Jaffe and H.M. Burte, Ed., Plenum Press, New York, 1973, p 2459-2474.

Ti-6Al-2Cb-1Ta-0.8Mo Titanium Alloy: Comparison of Sustained Load Crack Growth Data, in Several Environments

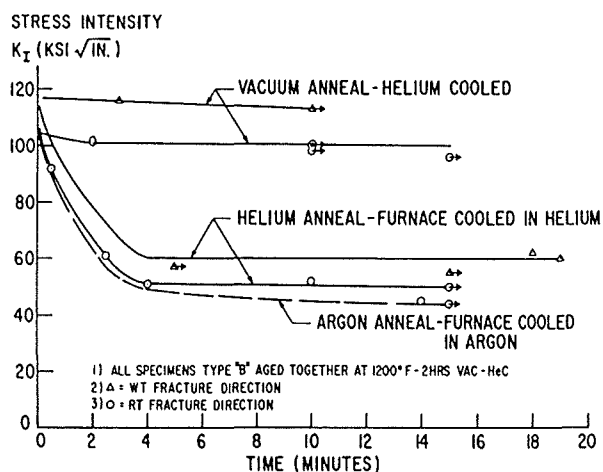


Comparison of sustained load crack growth data for Ti-6Al-2Cb-1Ta-0.8Mo (Ti-6211) tested in several environments.

The alloy had an α/β microstructure, had been rolled at 968 °C, and then heat treated at 704 °C for 2 h, and then air cooled. The initial hydrogen content was approximately 30 ppm.

Ti-7Al-1Mo-1V Titanium Alloy: Stress-Corrosion Cracking Resistance, in Vacuum, Argon, and in Helium

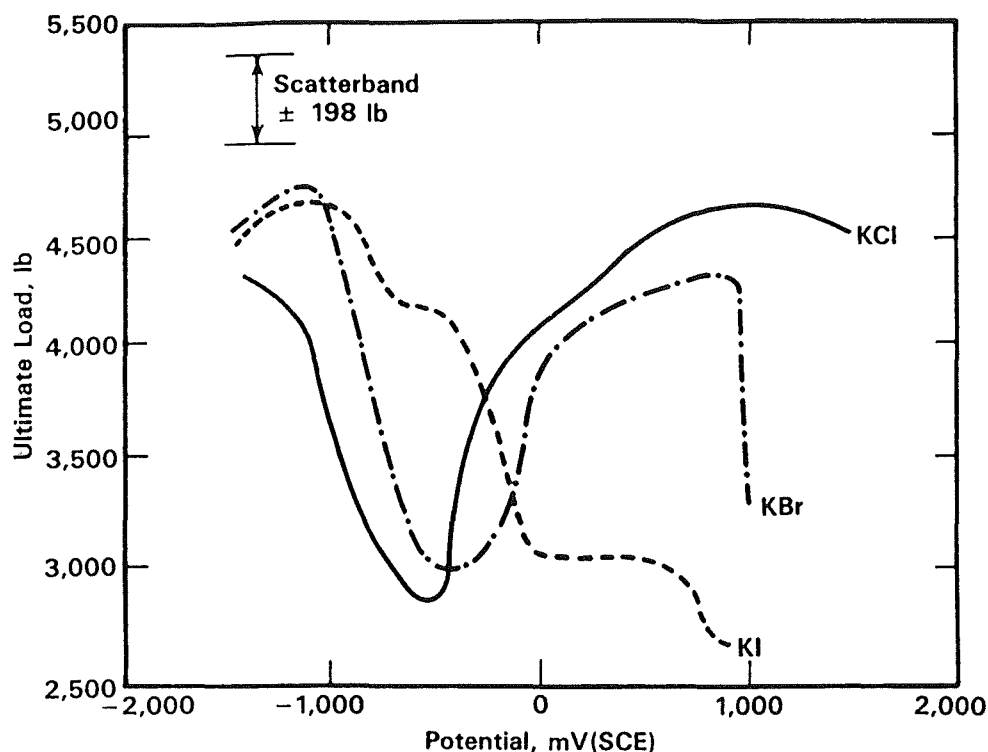
465



Stress-corrosion cracking resistance of Ti-7Al-1Mo-1V alloy solution annealed 1 h at 1700 °F in vacuum, argon, and helium followed by a 1200 °F age in vacuum and helium cooled. Tested in 2.5% sodium chloride solution.

Source: D.G. Howe and R.J. Goode, Effects of Heat Treating Environmental Conditions on the Stress-Corrosion Cracking Resistance of Several Titanium Alloys, in Applications Related Phenomena in Titanium Alloys, STP 432, American Society for Testing and Materials, Philadelphia, 1968, p 189-204.

Ti-8Al-1Mo-1V Titanium Alloy: Ultimate Fracture Load under Potentiostatic Conditions, in Halide Solutions

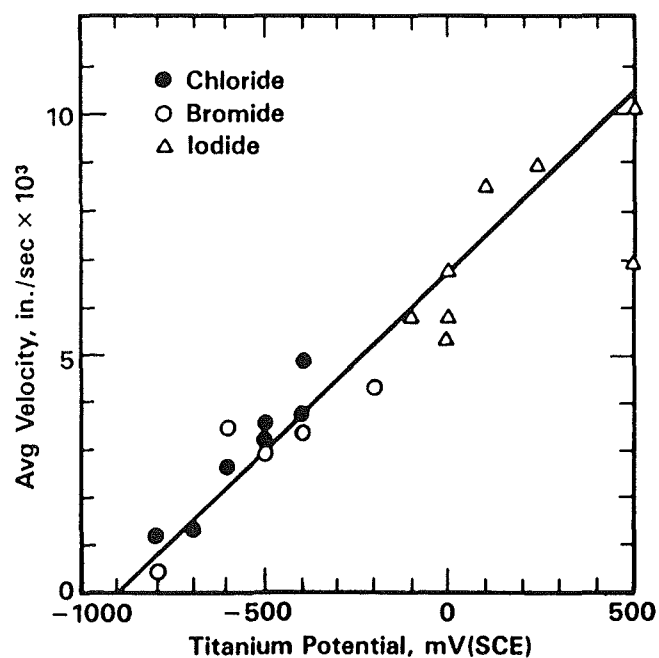


Ultimate fracture load of 0.050-in. thick duplex annealed Ti-8Al-1Mo-1V specimens in 0.6M halide solutions under potentiostatic conditions.

Effect of potential is important because titanium alloys are often coupled into other metals when incorporated into a structure. The minimum load at which fracture initiates in aqueous KCl solutions is around -500 mV_{SCE}, which is more noble than the free corrosion potential of titanium alloys in 3.5% sodium chloride solution and in seawater (-800 mV_{SCE}).

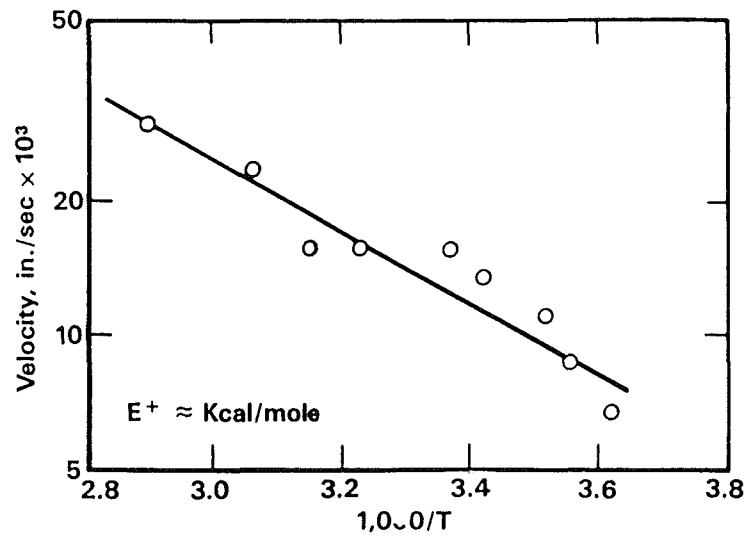
Ti-8Al-1Mo-1V Titanium Alloy: Relationship of Stress-Corrosion Cracking Velocity to Potential, in Neutral Halide Solutions

467



Relationship of stress-corrosion cracking plateau velocity to potential for Ti-8Al-1Mo-1V in 0.6M neutral halide solutions.

Ti-8Al-1Mo-1V Titanium Alloy: Relationship of Stress-Corrosion Cracking Plateau Velocity to Temperature, in KCl



Relationship of stress-corrosion cracking plateau velocity to temperature for Ti-8Al-1Mo-1V in 0.6M KCl at -500 mV.

Ti-8Al-Mo-1V Titanium Alloy: Effect of Alloy Composition on Salt-Stress-Corrosion Threshold

469

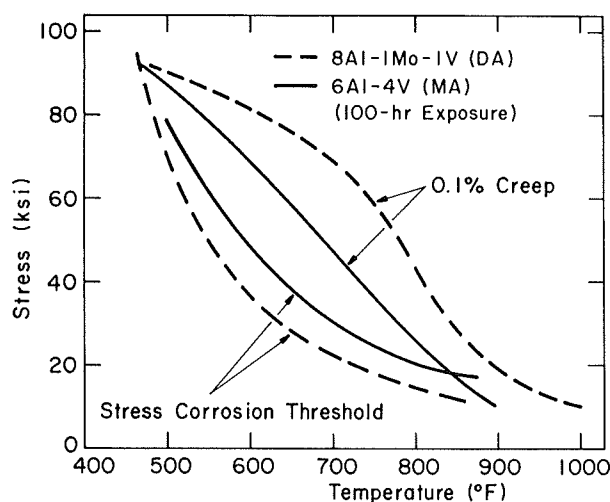


Table I—Effect of salt composition in the hot-salt stress-corrosion cracking of Ti alloys

Most Aggressive	M.P. °F	Least Aggressive	M.P. °F
LiCl-KCl	670	CaCl ₂	1420
LiCl	1138	CaCl	1195
NaCl	1474	SrCl ₂	1604
AgCl	850	MgCl ₂	1308
NaBr	1390	NaF	1812
NaI	1204		
KCl	1428		

Table II—Probable reactions in the hot-salt stress-corrosion of titanium alloys

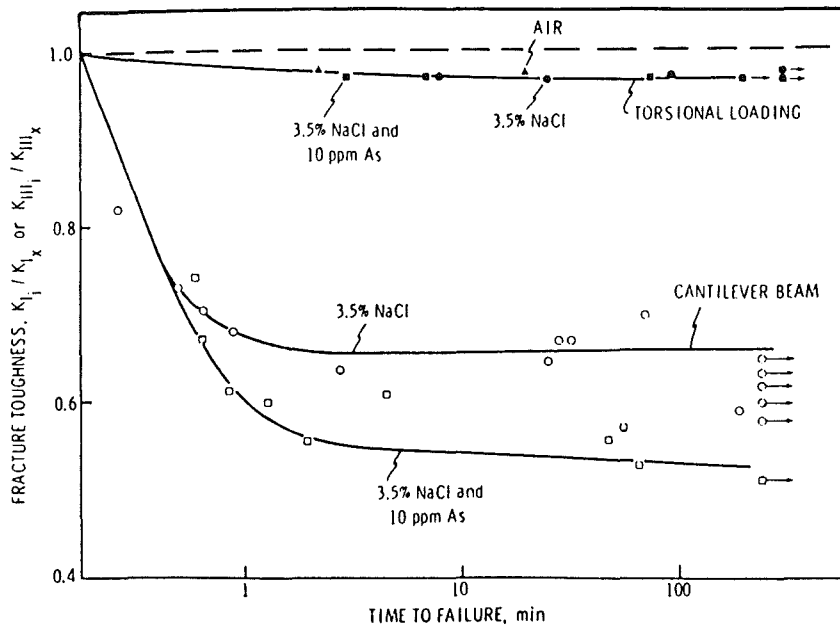
- (1) $\text{NaCl} + \text{H}_2\text{O} \rightarrow \text{HCl} + \text{NaOH}$
- (2) $\text{Ti} + 2\text{HCl} \rightarrow \text{TiCl}_2 + 2\text{H}$
- (3) $\text{Ti} + 2\text{NaCl} + \frac{1}{2}\text{O}_2 + \text{H}_2\text{O} \rightarrow \text{TiCl}_2 + 2\text{NaOH}$
- (4) $\text{TiCl}_2 + \text{H}_2\text{O} \rightarrow \text{TiO} + 2\text{HCl}$
- (5) $\text{TiCl}_2 + 2\text{H}_2\text{O} \rightarrow \text{TiO}_2 + 2\text{HCl} + 2\text{H}$
- (6) $\text{TiCl}_2 + \text{H}_2\text{O} \rightarrow \text{TiOCl}_2 + 2\text{H}$
- (7) $\text{Al} + 3\text{NaCl} + 3\text{H}_2\text{O} \rightarrow \text{AlCl}_3 + 3\text{NaOH} + 3\text{H}$
- (8) $2\text{AlCl}_3 + 3\text{H}_2\text{O} \rightarrow \text{Al}_2\text{O}_3 + 6\text{HCl}$

Effect of alloy composition on salt-stress-corrosion threshold.

In the 500 to 800 °F range, the stresses to produce salt cracking are well below the 0.1% creep design criterion. At about 900 °F, the creep and hot salt curves again merge. Commercially pure grades of titanium are immune to hot salt cracking, but all moderately alloyed grades are susceptible. The least resistant alloys are the all-alpha and near-alpha types, which are high in aluminum. Molybdenum appears to improve resistance to hot salt cracking. The heat treatable beta and alpha-beta alloys are more resistant.

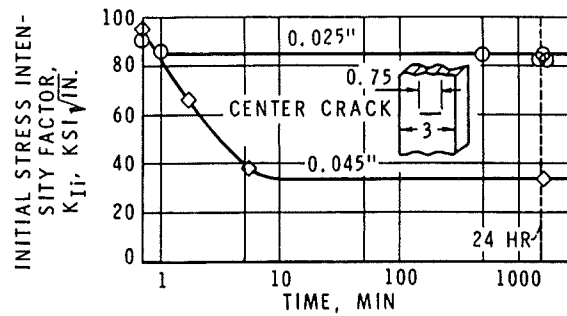
Possible reactions involved in hot salt cracking are given in Table II. Proposed cracking mechanisms include preferential attack of aluminum by HCl, hydride embrittlement, hydrogen embrittlement, and stress-sorption mechanisms.

Ti-8Al-1Mo-1V Titanium Alloy: Resistance to Stress-Corrosion Cracking as a Function of Bending or Torsion, in NaCl



Resistance to stress-corrosion cracking of Ti-8Al-1Mo-1V as a function of either bending or torsion loading mode.

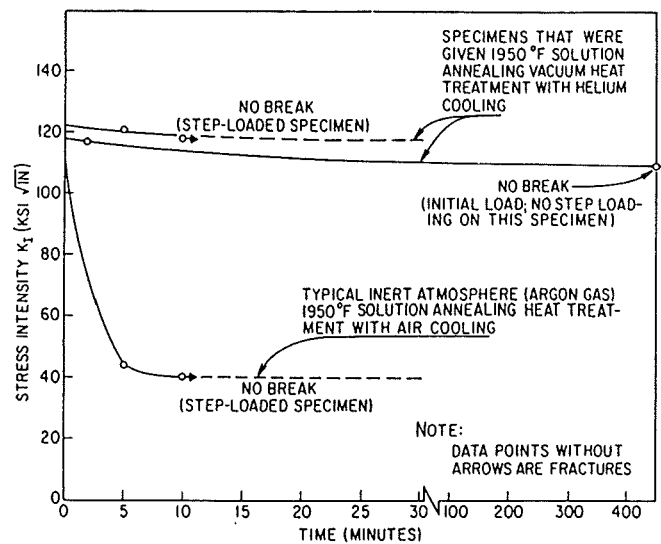
Arsenic additions were used to facilitate hydrogen entry and were carried out at $-500 \text{ mV}_{\text{SCE}}$. The lower resistance in Mode I to stress-corrosion cracking was taken as an indication that the hydrogen played an important role in the cracking process, because in Mode I a state of hydrostatic tensile stress exists ahead of the crack tip, which was thought to promote the concentration of hydrogen.



Stress-corrosion cracking results at room temperature for two thicknesses of Ti-8Al-1Mo-1V sheet tested in 3.5% sodium chloride solution.

The results illustrate that the susceptibility to stress-corrosion cracking is much greater under plane-strain rather than plane-stress conditions.

Ti-8Al-1Mo-1V Titanium Alloy: Stress-Intensity Versus Time-to-Failure, in Sodium Chloride

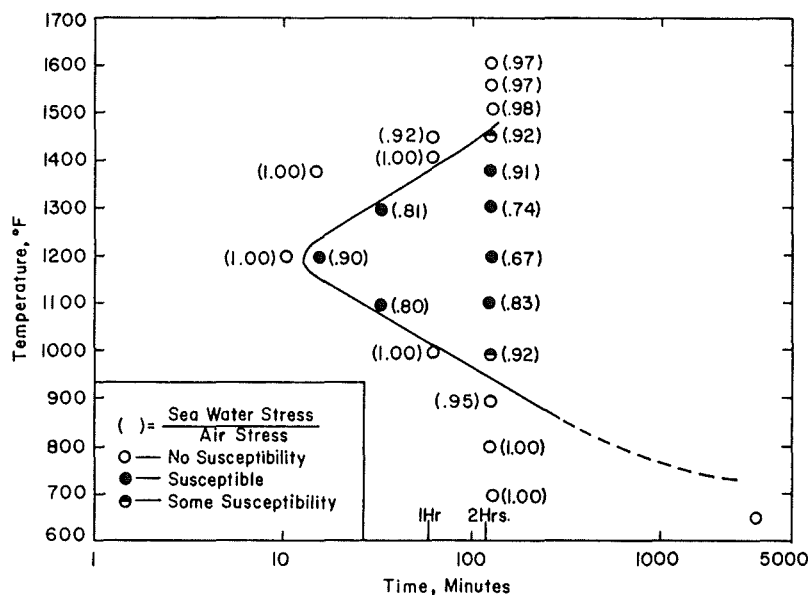


Stress-intensity versus time-to-failure curves for Ti-8Al-1Mo-1V in several heat treated conditions and tested in 3.5% sodium chloride solution.

Source: D.G. Howe and R.J. Goode, Effects of Heat Treating Environmental Conditions on the Stress-Corrosion Cracking Resistance of Several Titanium Alloys, in Applications Related Phenomena in Titanium Alloys, STP 432, American Society for Testing and Materials, Philadelphia, 1968, p 189-204.

Ti-8Al-1Mo-1V Titanium Alloy: Effect of Heat Treating Time and Temperature on Stress-Corrosion of Air Cooled Ti-811 Plate in Seawater

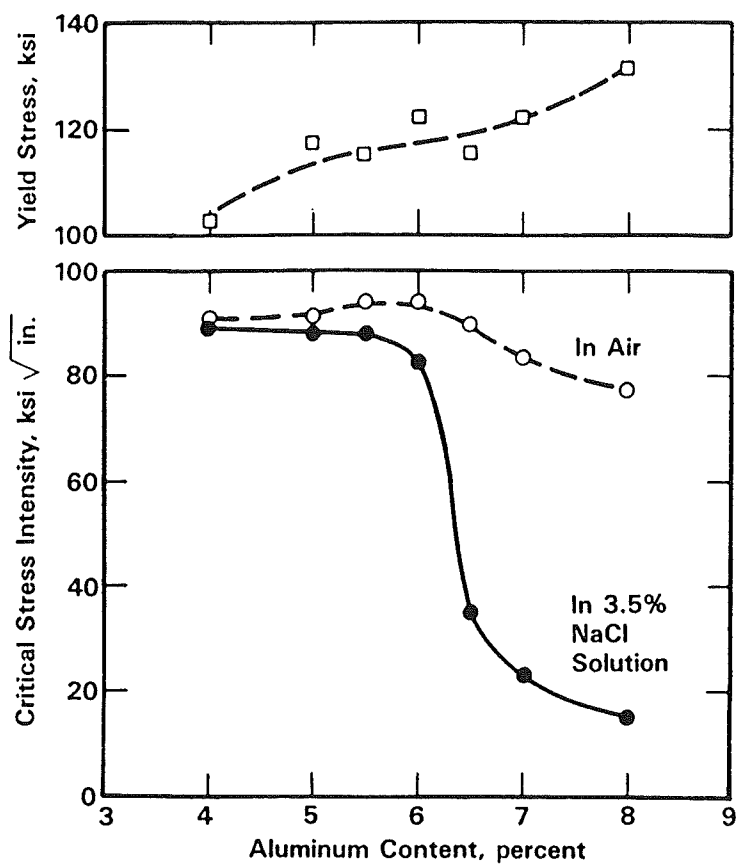
473



Effect of heat treating time and temperature on stress-corrosion incidents in Ti-8Al-1Mo-1V plate (air cooled).

The data show a time-temperature stress-corrosion relationship similar to metallurgical time-temperature-transformation diagrams. The most severe exposure and knee of the curve was at 1200 °F. The upper boundary for stress-corrosion appears to be near 1500 °F, the estimated Ti_3Al transus for the aluminum content. Once Ti_3Al forms, it is difficult to eliminate by thermal treatment at lower temperatures. The stress-corrosion resistance of titanium alloys containing aluminum is improved if the Ti_3Al phase is not present.

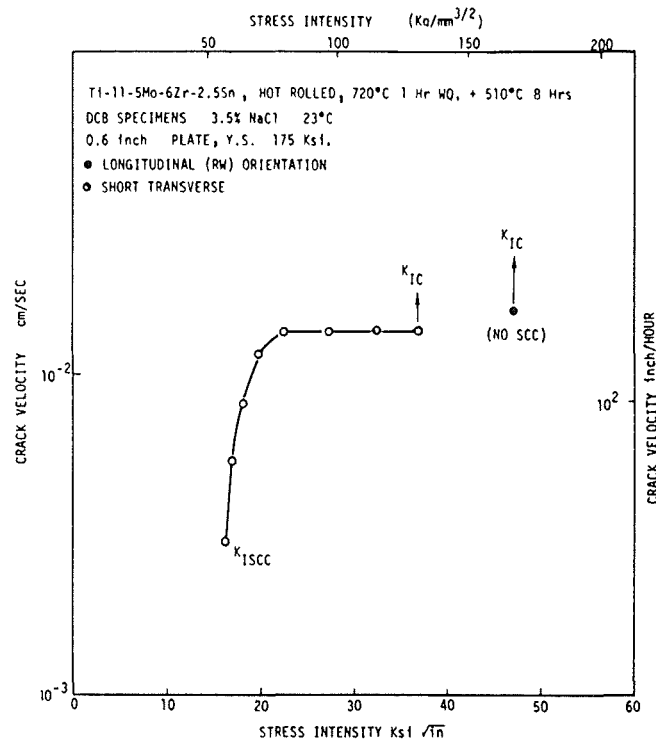
Ti-1.5Mo-5V Titanium Alloy: Effect of Aluminum Content on Strength and K_{Isc}



Effect of aluminum content on the strength, K_{Ic} , and K_{Isc} of a Ti-1.5Mo-5V alloy.

Ti-11.5Mo-6Zr-2.5Sn Titanium Alloy: Stress-Corrosion Cracking, in Sodium Chloride

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Stress-corrosion cracking behavior of Ti-11.5Mo-6Zr-2.5Sn tested in 3.5% sodium chloride solution.

The alloy was hot rolled and then heat treated at 720 °C for 1 h, water quenched, plus 510 °C for 8 h, and air cooled. Note that the material is much more susceptible to stress-corrosion cracking in the short-transverse direction.

**Titanium Alloys: Stress-Corrosion Cracking Resistance,
in Seawater and in Aqueous Sodium Chloride**

**TITANIUM ALLOYS WITH SCC RESISTANCE
IN SEAWATER AND AQUEOUS SODIUM
CHLORIDE SOLUTIONS^(50,51)**

Ti with O <0.3 weight percent

Ti-2Al-4Mo-4Zr

Ti-4Al-3Mo-1V

Ti-5Al-2Sn-2Mo-2V

Ti-6Al-2Mo

Ti-6Al-2Nb-Ta-0.8Mo

Ti-6Al-2Sn-1Mo-1V

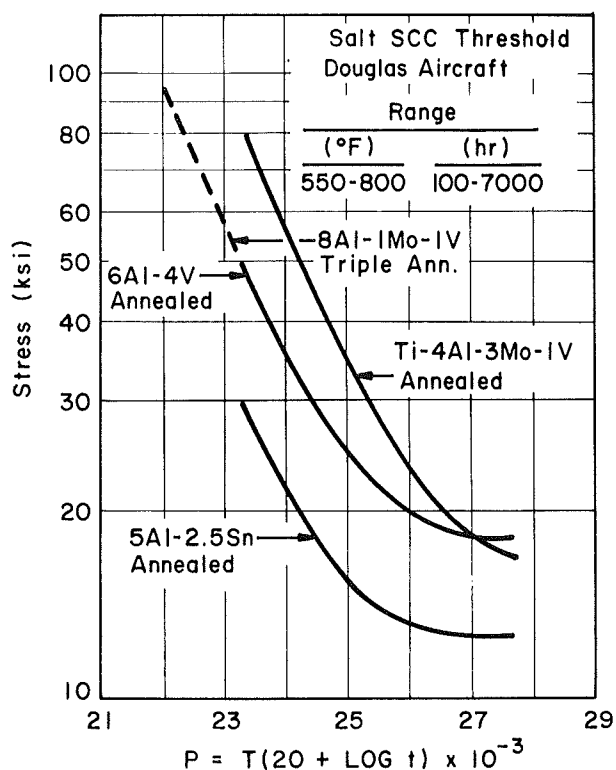
Ti-6Al-2Sn-1Mo-3V

Ti-6.5Al-5Zr-1V

Ti-7Al-2.5Mo (as received and beta annealed + WQ + 1100 F age for 2h)

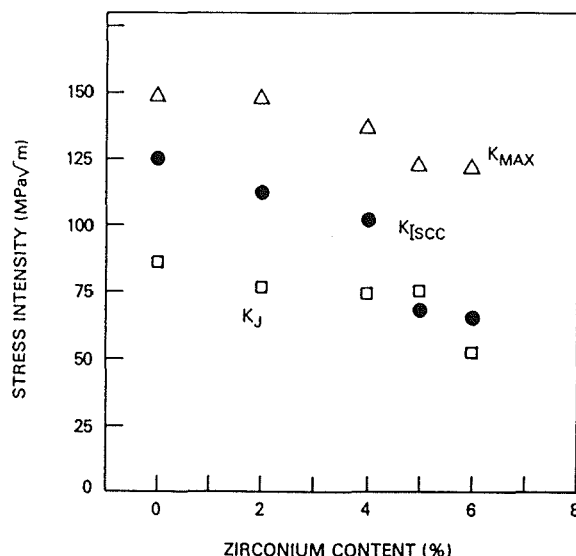
Ti-11.5Mo-6Zr-3.4Sn

Titanium alloys with stress-corrosion cracking resistance in
seawater and aqueous sodium chloride solutions.



Application of Larson-Miller parameter to threshold salt-stress-corrosion cracking in titanium alloys.

Titanium Alloys: Effect of Zirconium Content on K_{Isc} Values, in 3.5% NaCl Solution



Effect of zirconium content on K_{Isc} values of titanium alloys. Tested in 3.5% sodium chloride solution. Alloys: Ti-6Al-2V-xZr.

The alloys tested and their compositions (%) were:

Alloy	Al	V	Zr	O	H (ppm)
Ti-6Al-2V	6.00	1.95	...	0.061	43
Ti-6Al-2V-2Zr	5.75	1.80	1.95	0.062	48
Ti-6Al-2V-4Zr	5.95	1.95	4.05	0.056	33
Ti-6Al-2V-5Zr	6.00	1.90	5.00	0.059	40
Ti-6Al-2V-6Zr	6.00	2.00	6.05	0.054	46

K_J and K_{max} are computed from J-integral test results. K_J is the stress-intensity factor computed at the point of crack initiation, and K_{max} is the stress-intensity factor computed at maximum load. It was concluded that the low resistance to stress-corrosion cracking of alloys containing 5% or more of zirconium precludes their use in salt water environments.

The most significant reason for the good fracture toughness of these alloys is the low oxygen content. Oxygen content is probably the strongest determinant of all factors that impact mechanical properties. Oxygen content for all of the alloys was approximately 0.06 wt%, which is very low. Both static and dynamic fracture resistance were dramatically reduced by the addition of zirconium. Stress-corrosion cracking alloys with 5 and 6% zirconium were sensitive to stress-corrosion cracking in a 3.5% salt water solution, whereas alloys with less zirconium were immune to stress-corrosion cracking. It is noted that stress-corrosion cracking sensitivity also depends strongly on microstructure and that materials processed differently might exhibit different behavior.

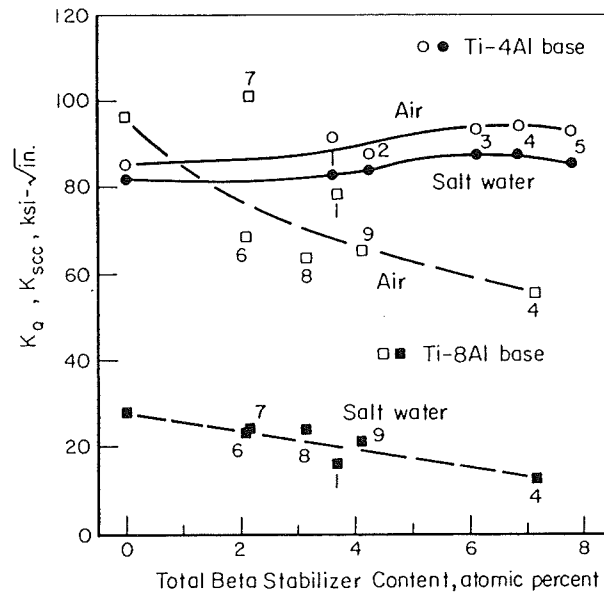
Titanium Alloys: Effect of Total Beta-Stabilizer Content on K_Q and K_{SCC}

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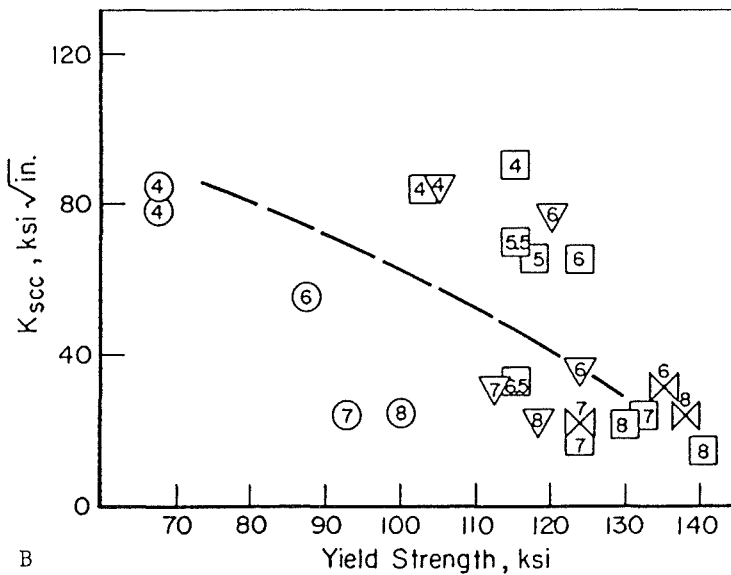
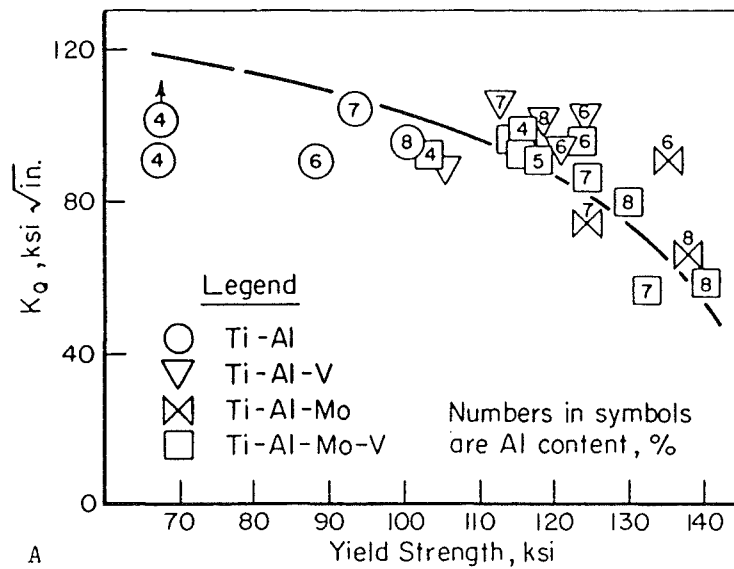
EFFECT OF BETA-STABILIZER CONTENT ON K_Q AND K_{SCC}

Nominal Composition, weight percent	K_Q , ksi√in.	K_{SCC} , ksi√in.	Yield Strength, ksi
Ti-4Al	> 100 (86)	77 (82)	67
Ti-4Al-1.5Mo-.5V	92	83	102
Ti-4Al-4V	88	85	106
Ti-4Al-2Mo-2V	94	88	108
Ti-4Al-3Mo-1V	95	88	114
Ti-4Al-4Mo	93	86	111
Ti-6Al	94	55	88
Ti-6Al-2V	102	34	123
Ti-6Al-1.5Mo-.5V	94	64	122
Ti-6Al-4V	92	74	121
Ti-6Al-2Mo	90	32	134
Ti-7Al	92	26	92
Ti-7Al-2V	108	28	111
Ti-7Al-1.5Mo-.5V	84	20	123
Ti-7Al-1.5Mo-.5V-2Pd	93	21	120
Ti-7Al-2Mo	72	21	123
Ti-7Al-3Mo-1V	56	19	134
Ti-7Al-2Mn	61	15	124
Ti-8Al	96	28	100
Ti-8Al-2V	102	24	118
Ti-8Al-1Mo	69	23	119
Ti-8Al-1Mo-1V	64	24	136
Ti-8Al-1.5Mo-.5V	79	15	130
Ti-8Al-2Mo	65	21	138
Ti-8Al-3Mo-1V	55	13	140

Titanium Alloys: Effect of Total Beta-Stabilizer Content on K_Q and K_{SCC} (Continued)



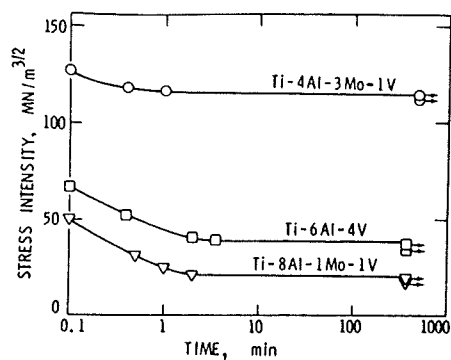
Effect of total β -stabilizer content on K_Q and K_{SCC} for Ti-4Al and Ti-8Al alloys. 1 = 1.5Mo-0.5V, 2 = 4V, 3 = 2Mo-2V, 4 = 3Mo-1V, 5 = 4Mo, 6 = 1Mo, 7 = 2V, 8 = 1Mo-1V, 9 = 2Mo.



Effect of aluminum content on K_Q and K_{SCC} at various strength levels. (a) Tests in air. (b) Tests in salt water.

Source: R.A. Wood, J.D. Boyd, and R.I. Jaffe, The Effects of Alloy Composition on the Salt-Water Stress-Corrosion Susceptibility of Titanium-Aluminum-Base Alloys, in Titanium Science and Technology, R.I. Jaffe and H.M. Burte, Ed., Plenum Press, New York, 1973, p 2639-2654.

Alpha + Beta Titanium Alloys: Time-to-Failure of Precracked Specimens, in Sodium Chloride

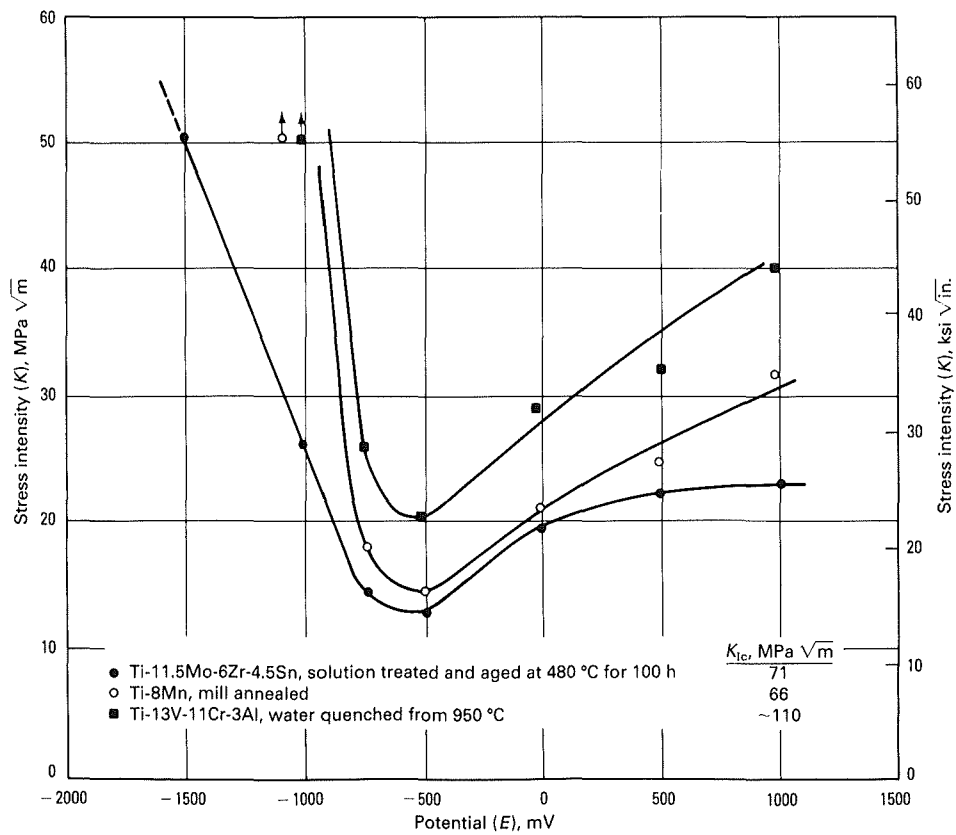


Time-to-failure of precracked specimens of three commercial α + β titanium alloys tested in 3.5% sodium chloride solution. All materials were in the mill annealed condition.

Note that the finite life portion in these tests is only a matter of minutes. The results also indicate that decreasing amounts of aluminum, particularly below 5%, or increasing amounts of molybdenum plus vanadium, increase the resistance to stress-corrosion cracking. Other factors such as oxygen content, the relative amounts of the α and β phases, and the fracture toughness may also play a role in determining resistance to stress-corrosion cracking.

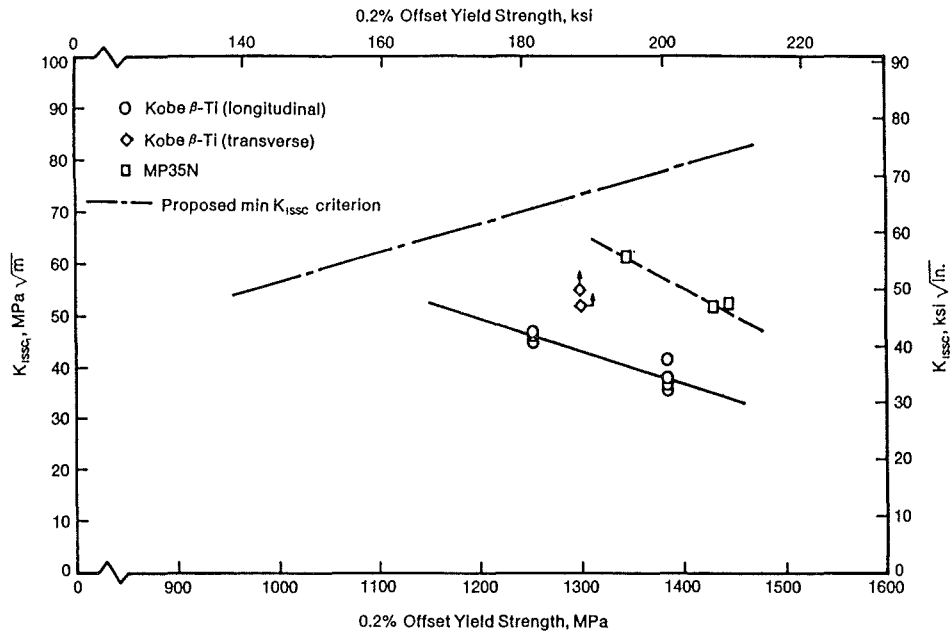
Beta-Titanium Alloys: Stress-Corrosion Cracking as a Function of Potential, in KCl

483

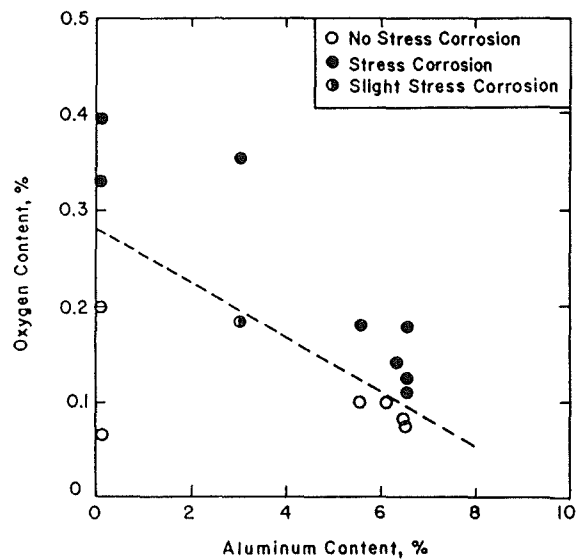


Stress-corrosion cracking of β -titanium alloys as a function of potential in 0.6M KCl at 24 °C.

Beta Titanium Alloy: Relationship Between Yield Strength and K_{ISSC} for Sulfide Cracking



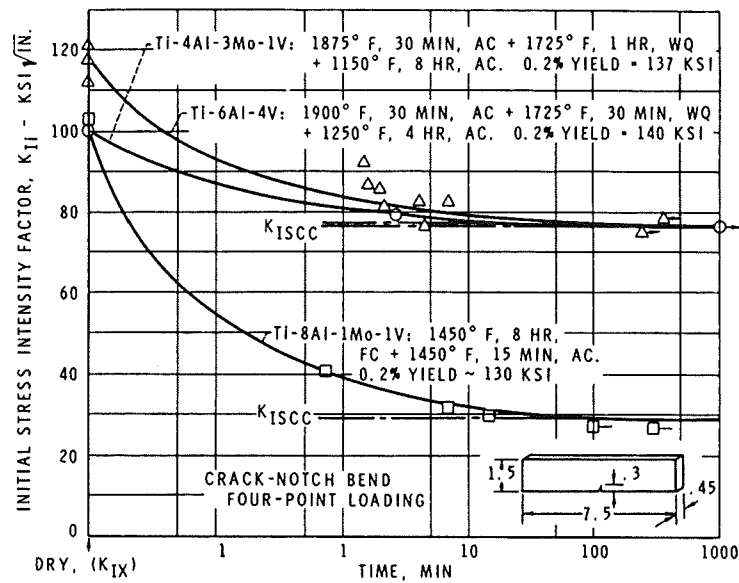
Relationship between yield strength and K_{ISSC} for sulfide cracking for a beta titanium alloy and MP35N (35Co-35Ni-20Cr-10Mo).



Influence of oxygen and aluminum content on stress-corrosion incidents.

As the aluminum content increases, less oxygen can be tolerated for good stress-corrosion resistance.

Titanium Alloys: Stress-Corrosion Cracking, in Sodium Chloride

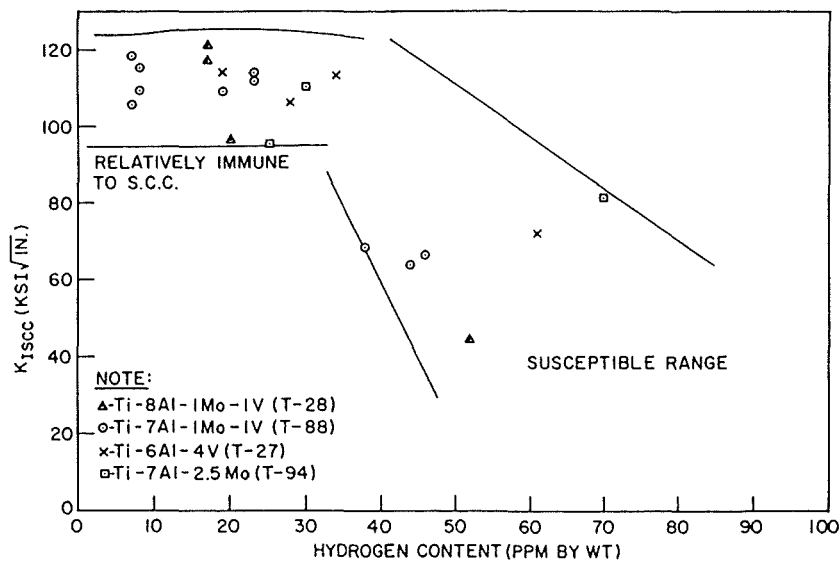


Stress-corrosion cracking results at room temperature for three titanium alloys tested in 3.5% sodium chloride solution.

Source: J.L. Shannon, Jr. and W.F. Brown, Jr., A Review of Factors Influencing the Crack Tolerance of Titanium Alloys, in Applications Related Phenomena in Titanium Alloys, STP 432, American Society for Testing and Materials, Philadelphia, 1968, p 33-63.

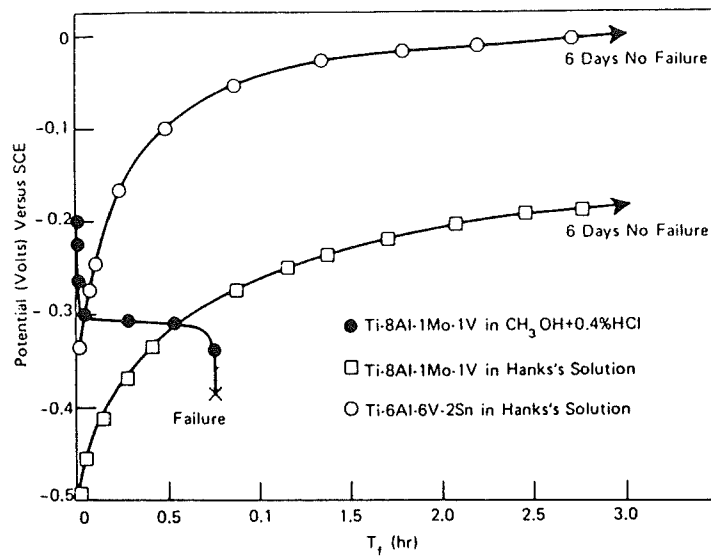
Titanium Alloys: Stress-Corrosion Cracking Resistance as Affected by Hydrogen Content, in Sodium Chloride

487



Stress-corrosion cracking resistance of several titanium alloys tested in 3.5% sodium chloride solution as affected by hydrogen content.

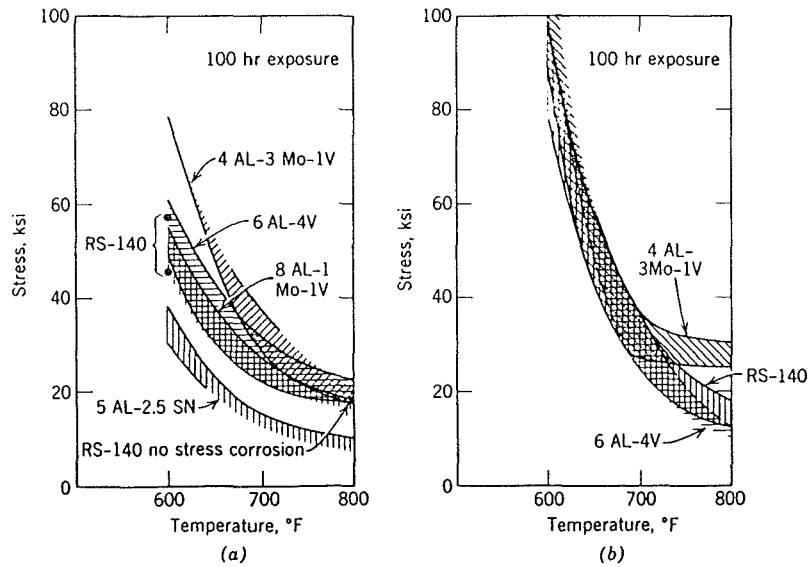
Titanium Alloys: Potential Time Curves, in $\text{CH}_3\text{OH} + \text{HCl}$ and in Hanks Solution



Potential time curves for titanium alloys stressed in $\text{CH}_3\text{OH} + 0.4\%\text{HCl}$ solution and in Hanks solution.

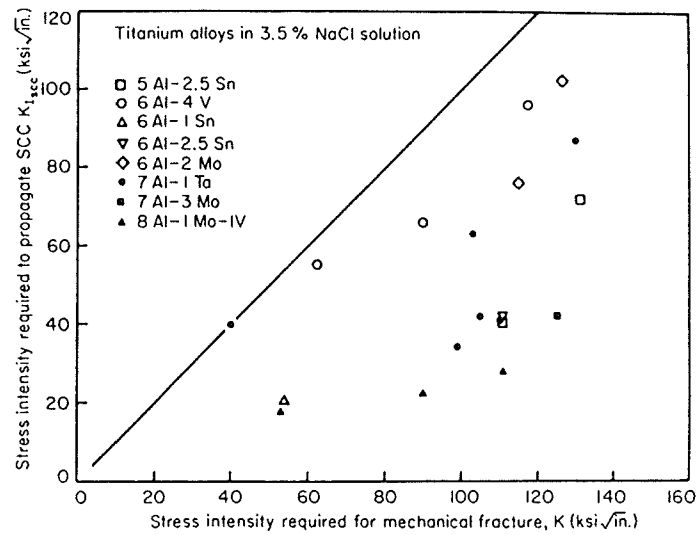
Titanium Alloys: Threshold Stress for Stress-Corrosion Cracking as a Function of Temperature

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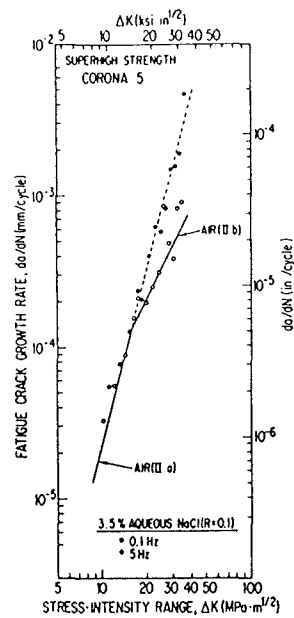


Threshold stress for stress-corrosion cracking of titanium alloys as a function of temperature. (a) Annealed alloys. (b) Solution treated and aged alloys.

Titanium Alloys: Lack of Correlation Between Threshold and Unstable Fracture Stress-Intensities, in Dilute Sodium Chloride

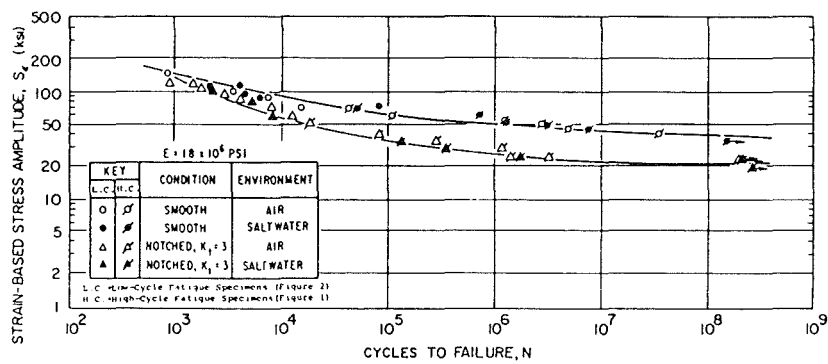


Lack of correlation between threshold and unstable fracture stress-intensities for titanium alloys. The stress-corrosion tests were in dilute sodium chloride solution.



Corrosion-fatigue crack growth rates in Corona 5 (Ti-4.5Al-5Mo-1.5Cr) tested in salt water at two different frequencies.

Ti-6Al-2Cb-1Ta-0.8Mo Titanium Alloy: Flexural Fatigue Curves



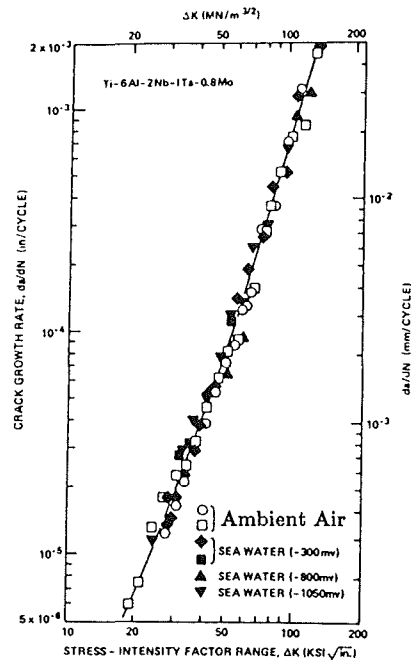
Flexural fatigue curves for Ti-6Al-2Cb-1Ta-0.8Mo.

The salt water environment has had little influence in these tests.

Source: R.C. Schwab and E.J. Czyryca, Effects of Notches and Saltwater Corrosion on the Flexural Fatigue Behavior of High-Strength Structural Alloys, in Effects of Environment and Complex Load History on Fatigue Life, STP 462, American Society for Testing and Materials, Philadelphia, 1970, p 203-216.

Ti-6Al-2Nb-1Ta-0.8Mo Titanium Alloys: Fatigue Crack Growth as a Function of Environment and Imposed Electrochemical Potential

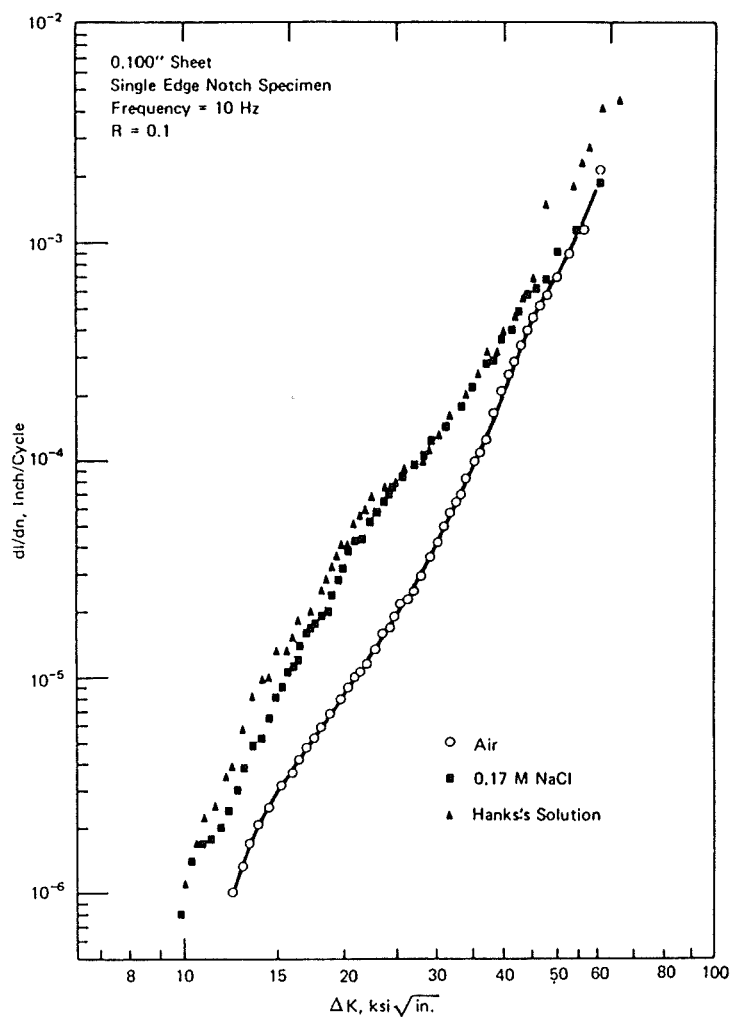
493



Fatigue crack growth behavior of Ti-6Al-2Nb-1Ta-0.8Mo as a function of environment and imposed electrochemical potential.

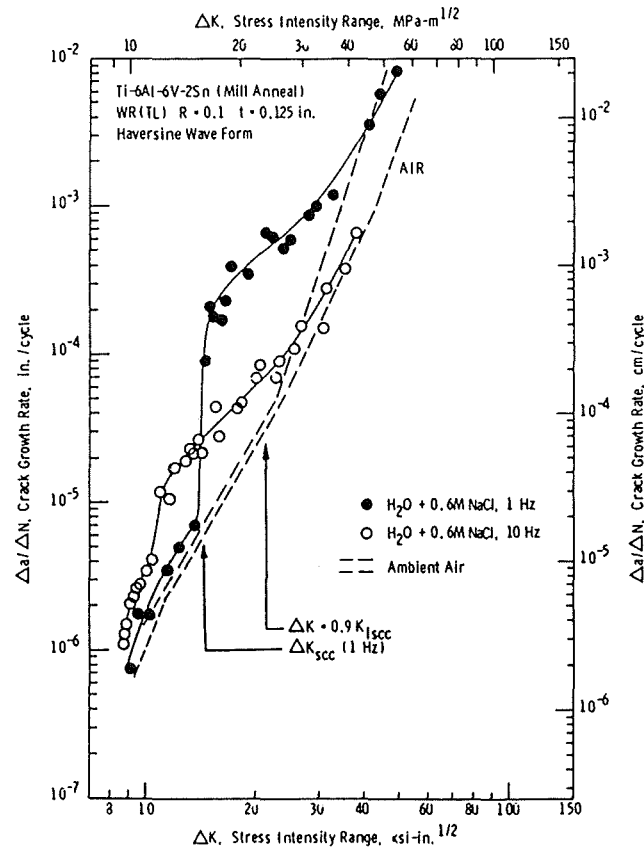
Source: C.R. Crowe and D.F. Hasson, The Use of Titanium in Deep Sea Exploration, in Titanium for Energy and Industrial Applications, D. Eylon, Ed., American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, PA, 1981, p 93-110.

Ti-6Al-6V-2Sn Titanium Alloy: Effect of Environment on Fatigue Crack Propagation



Effect of environment on fatigue crack propagation in Ti-6Al-6V-2Sn.

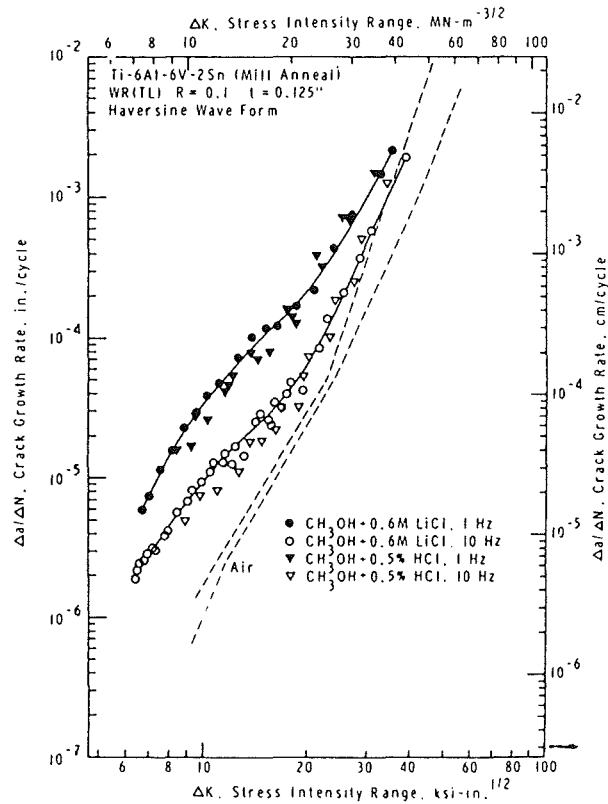
Source: M. Levy, et al., The Corrosion Behavior of Titanium Alloys in Chloride Solutions: Materials for Surgical Implants, in Titanium Science and Technology, R.I. Jaffe and H.M. Burte, Ed., Plenum Press, New York, 1973, p 2459-2474.



Effect of frequency on fatigue crack growth behavior of Ti-662 in a 0.6M sodium chloride aqueous environment.

ΔK denotes the ΔK level at which there is an abrupt increase in the rate of fatigue crack growth. Note that the frequency effect at high and low ΔK is reversed. At low Δ levels, it was proposed that at the lower frequency repassivation occurred, whereas at high Δ it did not. It is also noted that for a fatigue stress ratio $R = 0.1$, $K_{\max} \geq K_{Isc}$ for $\Delta K \geq 0.9 K_{Isc}$.

Ti-662 Titanium Alloy: Fatigue Crack Growth, in Methanol + LiCl and Methanol + HCl

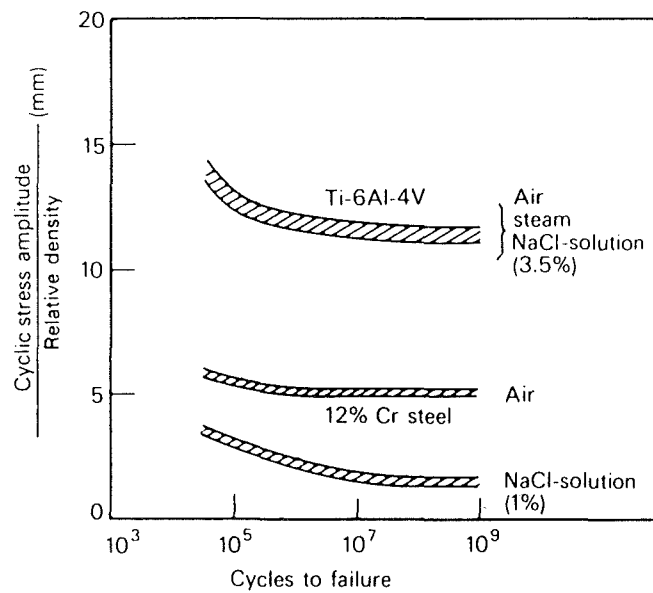


Fatigue crack growth behavior of Ti-662 at 1 and 10 Hz in methanol + 0.6M LiCl and in methanol + 0.5 vol% HCl.

In contrast to the previous figure, there is no reversal in frequency dependence at low ΔK levels. Repassivation, even at the lower frequency, did not occur.

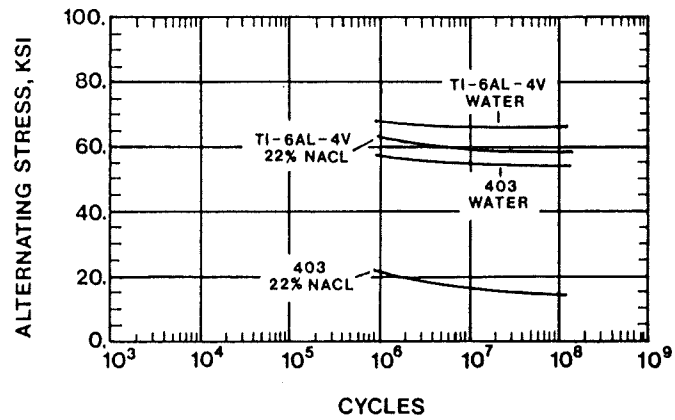
Ti-6Al-4V: Fatigue Properties Tested in Various Environments

497



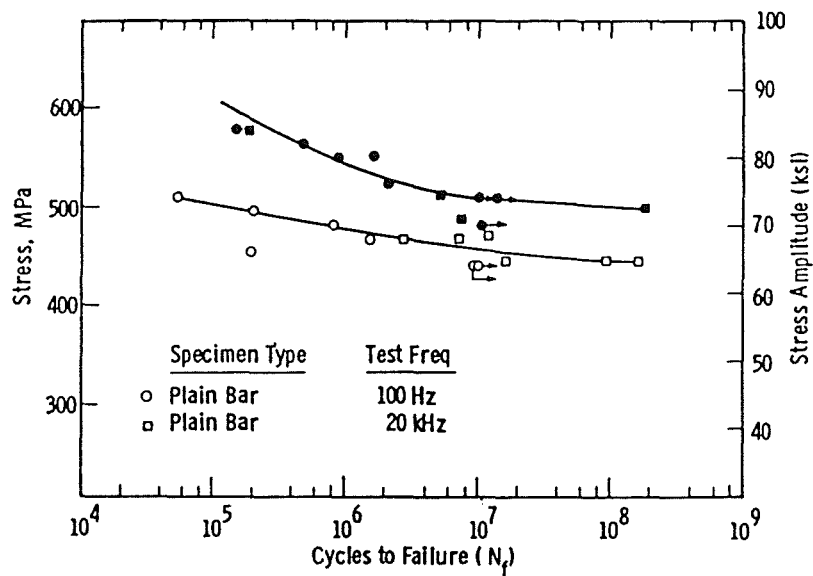
Fatigue properties (compensated for differences in relative density) of Ti-6Al-4V and a 12% chromium steel tested in various environments.

Ti-6Al-4V and Type 403 Stainless Steel: Comparison of Corrosion Fatigue Strength



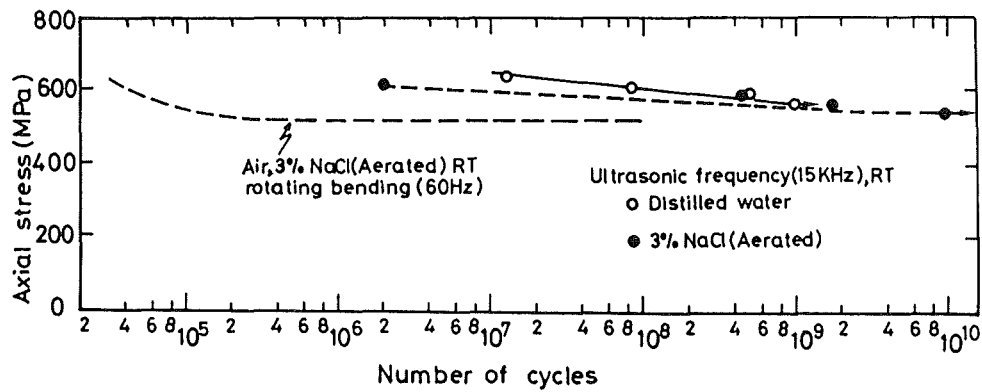
Corrosion fatigue strength comparison of Ti-6Al-4V and Type 403 stainless steel in 22% sodium chloride solution at 176 °F, with < 20 ppb O₂, pH 4.

Source: T.M. Ruse and W.G. Steltz, Titanium for Steam Turbine Blades, in Titanium for Energy and Industrial Applications, D. Eylon, Ed., American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, PA, 1981, p 11-28.



Comparison of conventional and ultrasonic fatigue strengths of fine-grained equiaxed α - β (open points) and fine-grained lamellar α - β (closed points) Ti-6Al-4V tested in pure water.

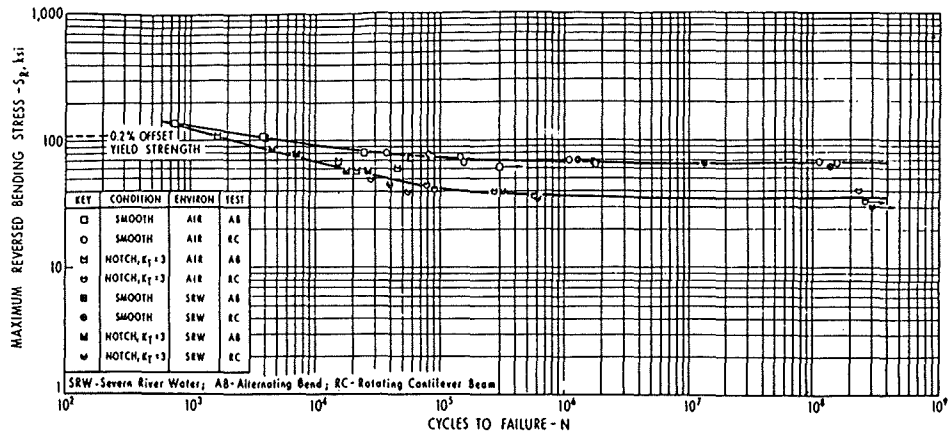
Ti-6Al-4V Titanium Alloy: Fatigue Curves, in Air and in Sodium Chloride



Fatigue curves for Ti-6Al-4V in air and in 3% sodium chloride solution at 60 Hz and 15 kHz test frequencies.

Ti-6Al-4V Titanium Alloy: Flexural Fatigue Curves, in Air and in Severn River Water

501

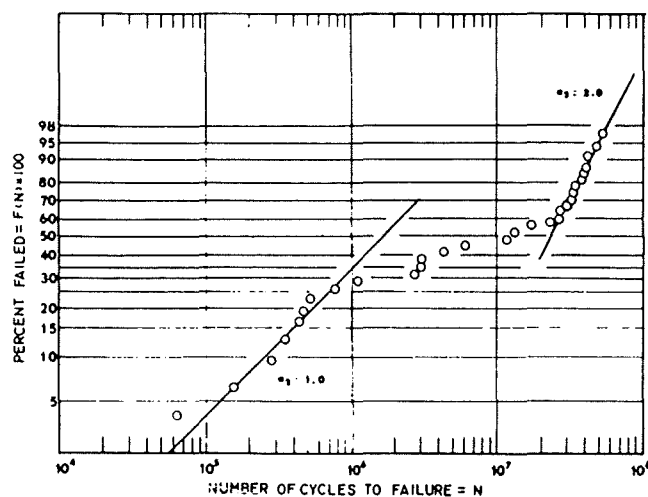


Flexural fatigue curves for Ti-6Al-4V containing 800 ppm oxygen maximum tested in air and in Severn River water. Severn River water is brackish, yet there is no discernible effect of the environment in these tests.

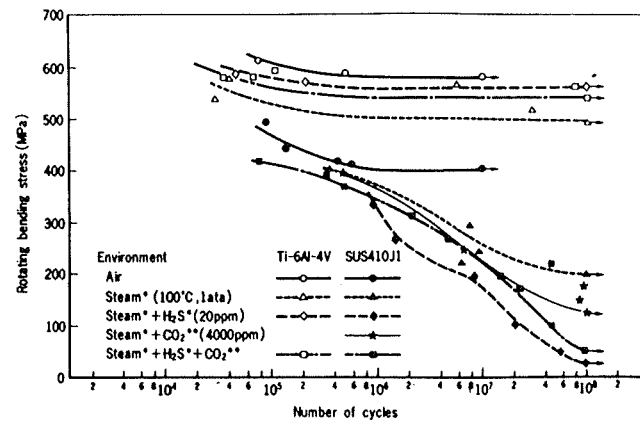
Source: N.G. Feige and R.L. Kane, Service Experience with Titanium Structures in Marine Service, in Proceedings of the 26th Conference, National Association of Corrosion Engineers, Houston, 1970, p 194-199.

Ti-6Al-4V Titanium Alloy: Weibull Distribution for High-Frequency Fatigue, in Water

Material: Titanium 6Al - 4V Annealed Stress Level: 52,800 psi (AXIAL PUSH PULL)
 Number of Specimens Tested: 30 Temperature: 70°F
 Frequency of Test: 14,000 c/s Environment: Water Cooling Bath

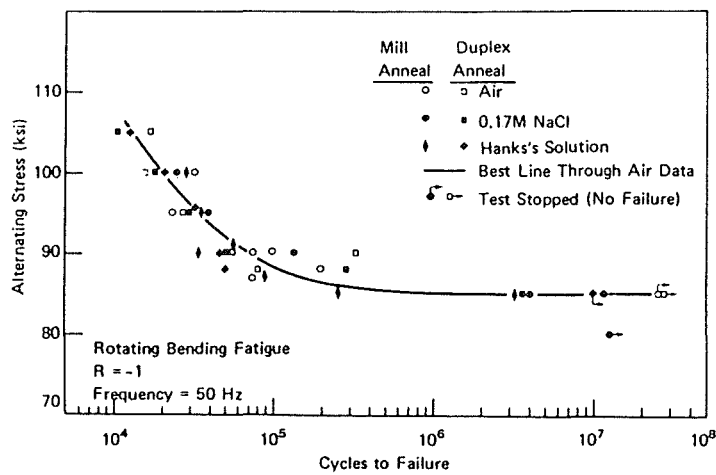


Weibull distribution for high-frequency fatigue of annealed Ti-6Al-4V in water.



Comparison of the effect of industrial gas in steam on the fatigue strength of Ti-6Al-4V and a stainless steel (SUS410J1).

Ti-8Al-1Mo-1V Titanium Alloy: Effect of Environment on Fatigue Life

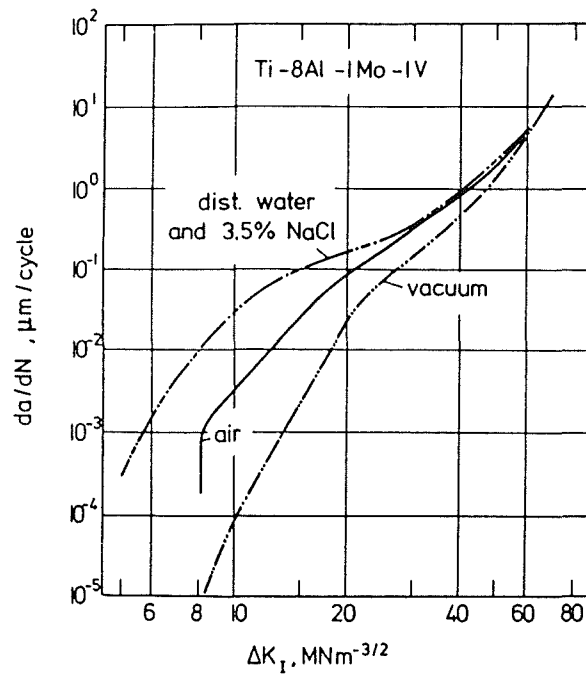


Effect of environment on the fatigue life of Ti-8Al-1Mo-1V.

Source: M. Levy, et al., The Corrosion Behavior of Titanium Alloys in Chloride Solutions: Materials for Surgical Implants, in Titanium Science and Technology, R.I. Jaffe and H.M. Burte, Ed., Plenum Press, New York, 1973, p 2459-2474.

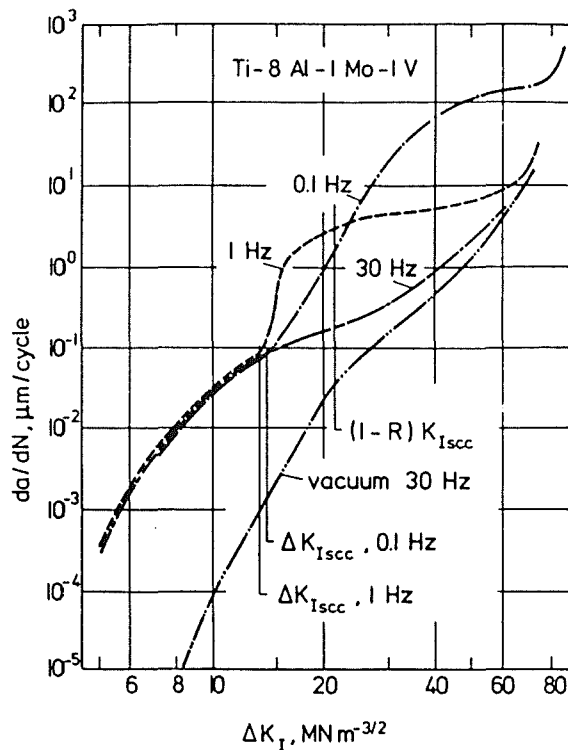
Ti-8Al-1Mo-1V Titanium Alloy: da/dN versus ΔK Curves, in Several Environments

505



da/dN versus ΔK curves for Ti-8Al-1Mo-1V in several environments at a frequency of 30 Hz.

Ti-8Al-1Mo-1V Titanium Alloy: Effect of Frequency on da/dN Versus ΔK Curves, in Sodium Chloride

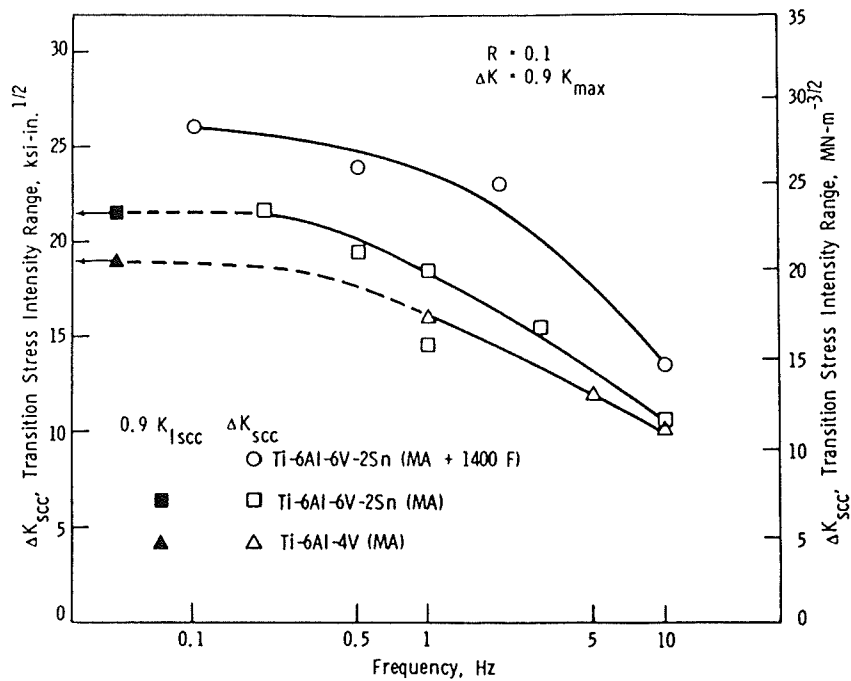


Effect of frequency on the da/dN versus ΔK curves for Ti-8Al-1Mo-1V in 3.5% sodium chloride solution.

All fatigue tests were run at $R = 0.1$ with sinusoidal waveform.

Titanium Alloys: Effect of Test Frequency on Transition Stress-Intensity Factor Range

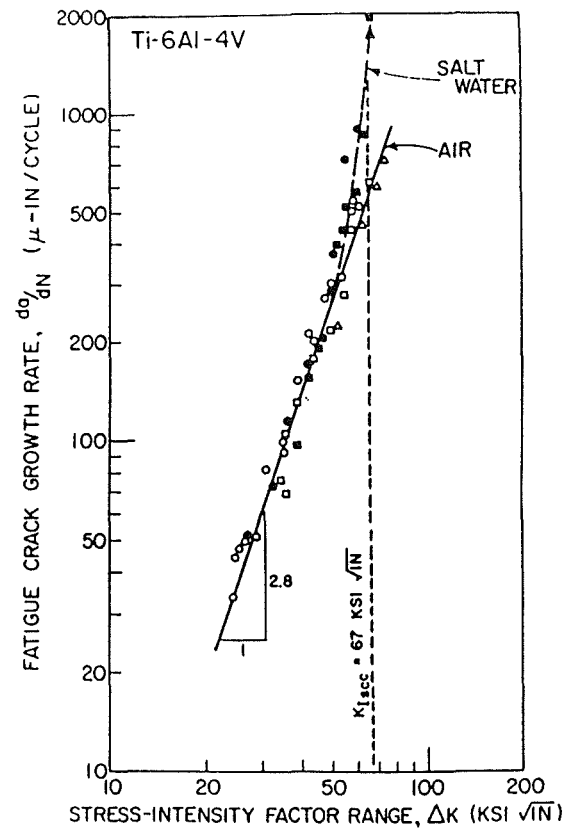
507



Effect of test frequency on ΔK_{SCC} , the transition stress-intensity factor range for cyclic stress-corrosion cracking.

ΔK level for which $K_{max} = 0.9 K_{Isc}$ is shown for reference. ΔK_{SCC} in fatigue is lower than K_{Isc} (the stress-corrosion cracking threshold under sustained loading), and it decreases with increasing frequency due to repeated rupture of the passive film at the crack tip.

Ti-6Al-4V Titanium Alloy: Fatigue Crack Growth, in Air and in Salt Water

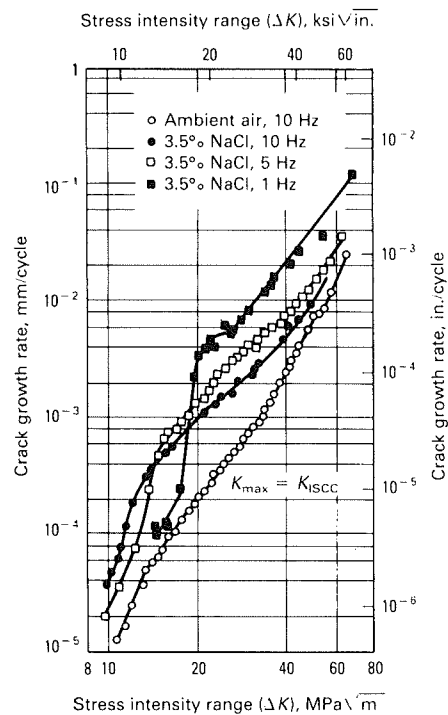


Fatigue crack growth behavior of Ti-6Al-4V in air and in salt water.

Note that there is little effect of the salt water environment below the K_{Isc} level.

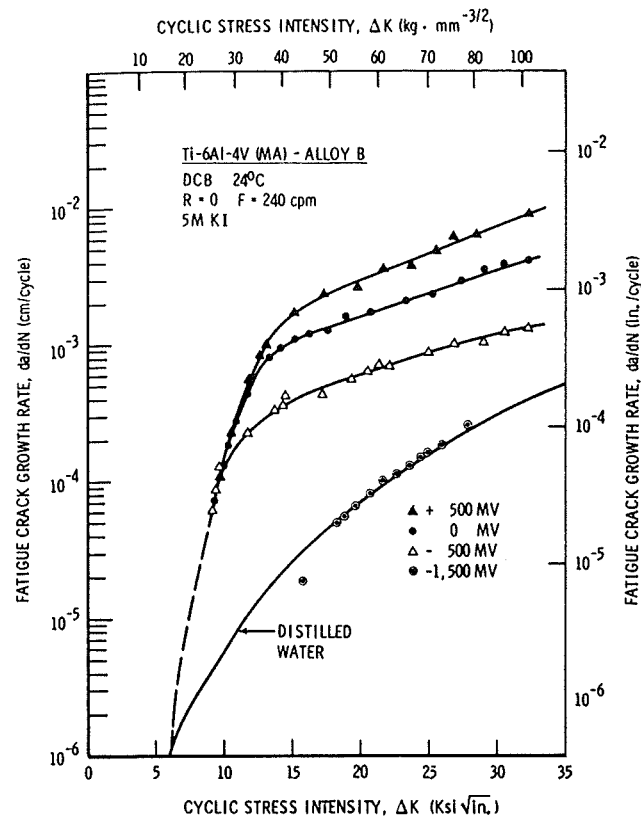
Ti-6Al-4V Titanium Alloy: Effects of Stress-Intensity, Frequency on Corrosion Fatigue Crack Growth, in Aqueous Sodium Chloride

509

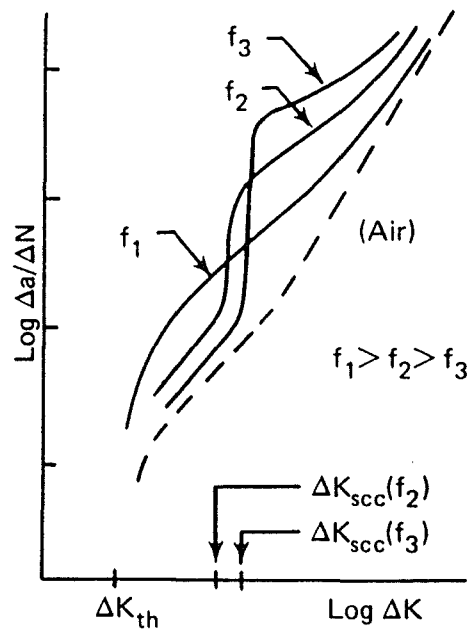


Effects of stress-intensity and frequency on corrosion fatigue crack growth in Ti-6Al-4V in aqueous sodium chloride at 23 °C. Stress ratio, $R = 0.1$.

Ti-6Al-4V Titanium Alloy: Effect of Potential Fatigue Crack Growth Rate, in Iodide Solutions and in Distilled Water



Effect of potential on fatigue crack growth rate in Ti-6Al-4V in iodide solutions. Note that at a potential of -1500 mV the crack growth rates are equal to those obtained in distilled water.



Effect of cyclic frequency on fatigue crack growth characteristics of titanium alloys in various environments.

Source: D.B. Dawson and R.M. Pelloux, Corrosion Fatigue Cracking of Titanium Alloys in Aqueous and Methanolic Environments, in 1972 Tri-Service Conference on Corrosion, MCIC 73-19, M.M. Jacobson and A. Gallaccio, Ed., Metals and Ceramics Information Center, Battelle, Columbus, 1972, p 77-94.

OTHERS

Mg, Ta, W, U, Zn, Zr

MAGNESIUM, TANTALUM, TUNGSTEN, URANIUM, ZINC, ZIRCONIUM

Magnesium alloys containing more than 1.5% aluminum are susceptible to stress-corrosion cracking (SCC). Although the incidence of SCC in service is fairly rare due to the low level of applied stress as well as to stress relaxation, cast alloys have undergone SCC in laboratory tests at stress levels of 50% of yield. In environments causing negligible corrosion, wrought alloys appear to be more susceptible than cast. Substantial reductions in fatigue strength have been found in laboratory tests using salt spray or drops.

Tantalum has undergone hydrogen assisted cracking in some severe aqueous acid media in chemical industry applications where it was electrically coupled to a less noble material—such as low carbon steel. Under these conditions, tantalum became the cathode in the galvanic cell thus created. Hydrogen embrittlement has been observed when tantalum was exposed to concentrated H_2SO_4 at 250°C or to concentrated HCL at 150°C.

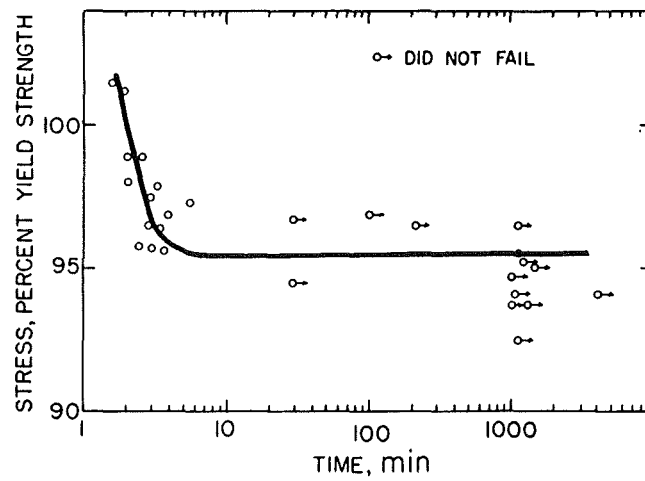
Uranium-titanium alloys, uranium-molybdenum alloys, uranium-niobium alloys, uranium-niobium-zirconium alloys and polynary uranium alloys containing small amounts of titanium, niobium, molybdenum, and zirconium, are susceptible to SCC in distilled water and in chloride solutions. Both transgranular as well as intergranular cracking paths have been observed.

Zinc is susceptible to SCC in tap water. Zinc coatings on steel are used to improve the corrosion fatigue resistance of the steel by having the zinc act as a sacrificial anode. However, in the case of high-strength steel, a break in the zinc coating can promote hydrogen embrittlement of the steel.

Zirconium and its alloys resist SCC in many environments—such as NaCl, HCl, MgCl_2 , NaOH, and H_2S —that would induce SCC in other alloys. Zirconium service failures are few in service applications. The high SCC resistance of zirconium can be attributed to its high repassivation rate. Any break in the oxide film will be quickly healed if sufficient oxygen is present. Even in dehydrated systems, sufficient oxygen is generally present for repassivation. The environments known to cause SCC of zirconium alloys include FeCl_3 or CuCl_2 solutions, mixtures of methanol (CH_3OH) and HCl or methanol and iodine, concentrated HNO_3 , and liquid mercury or cesium. A crystallographic texture such that the hexagonal basal planes are perpendicular to the tensile stress diminishes susceptibility to SCC.

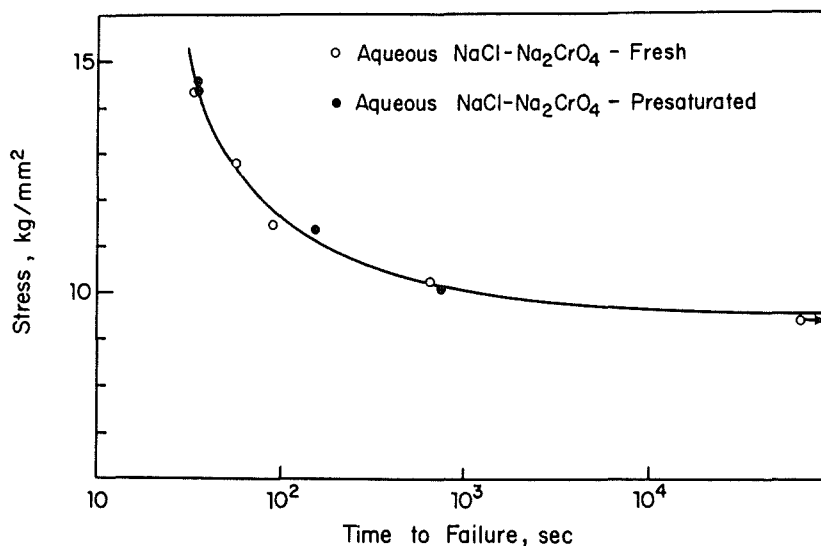
AZ31 Magnesium Alloy: Time-to-Failure, in NaCl + K₂CrO₄ Aqueous Solution

517



Time-to-failure for AZ31 (3Al-1Zn) magnesium alloy exposed in a 3.5% NaCl + 2% K₂CrO₄ aqueous solution at 30 °C.

AZ31 Magnesium Alloy: Time-to-Failure, in Fresh and Presaturated NaCl + Na₂CrO₄

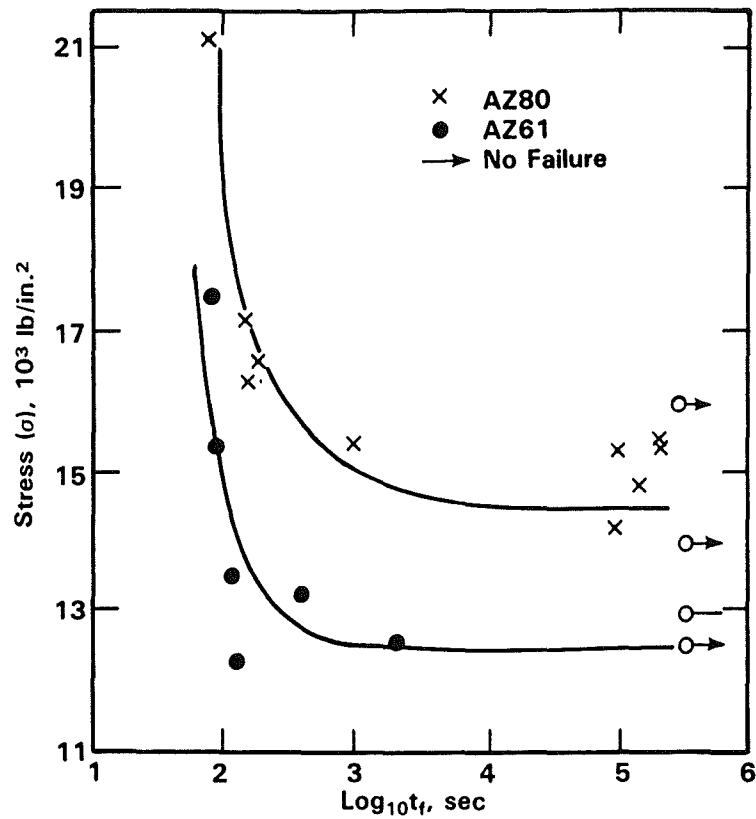


Time-to-failure of AZ31 magnesium alloy stressed in fresh and presaturated NaCl + Na₂CrO₄.

Tension specimens were tested over a range of engineering stresses in solutions containing 35 g/L sodium chloride and 30 g/L sodium chromate. Times-to-failure for fresh and presaturated solutions were not significantly different.

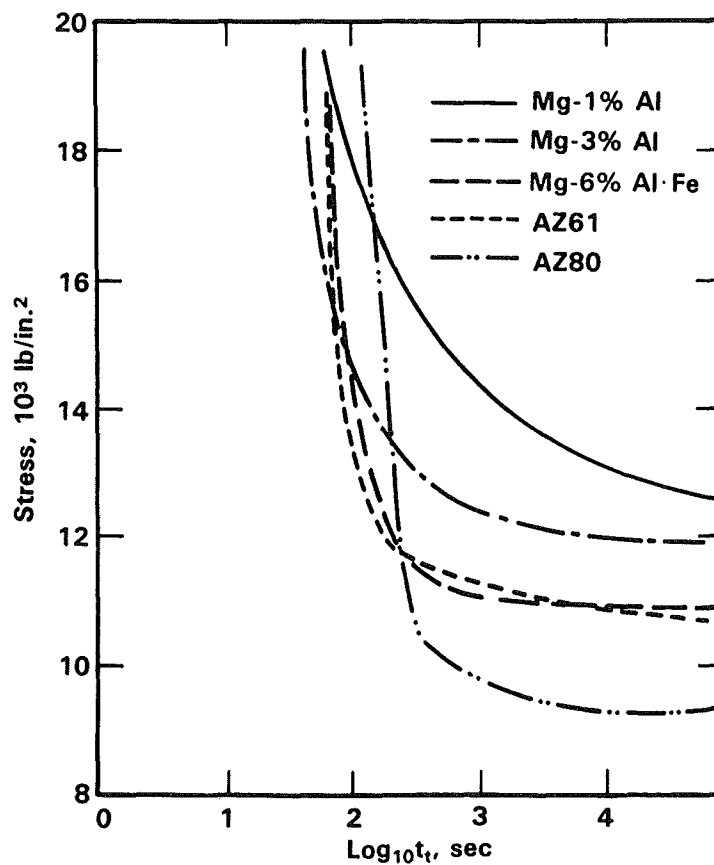
AZ80 and AZ61 Magnesium Alloys: Stress Versus Time-to-Failure, in Aqueous NaCl + Na₂CrO₄

519



Stress versus time-to-failure (t_f) for the two-phase alloys AZ80 (8.5Al-0.5Zn) and AZ61 (6Al-1Zn) in aqueous 40 g/L NaCl + 40 g/L Na₂CrO₄.

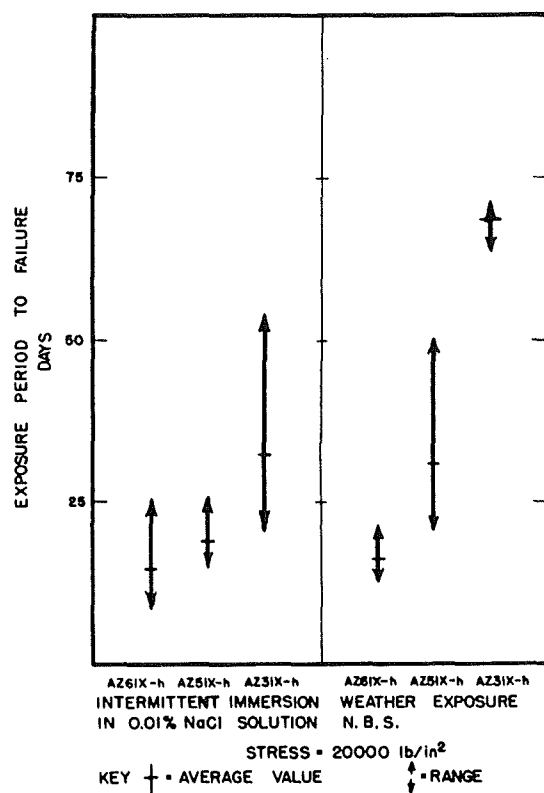
**Mg-Al Alloys: Stress Versus Time-to-Failure,
in Aqueous NaCl + Na₂CrO₄**



Stress versus time-to-failure (t_f) for magnesium-aluminum alloys in aqueous 40 g/L NaCl + 40 g/L Na₂CrO₄.

Magnesium Alloys: Comparison of Stress-Corrosion Cracking, in NaCl and After Weather Exposure

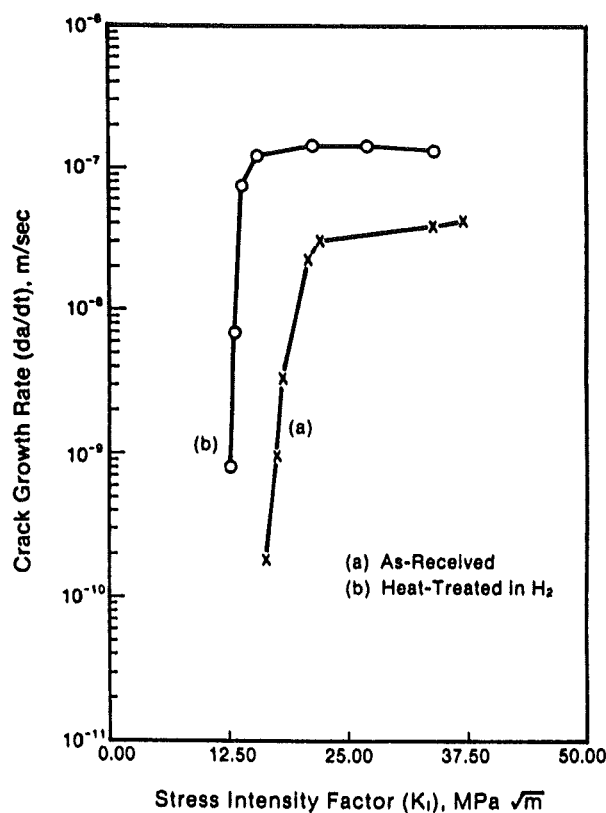
521



Comparison of exposure periods to failure of three magnesium alloys in the rolled (h) condition.

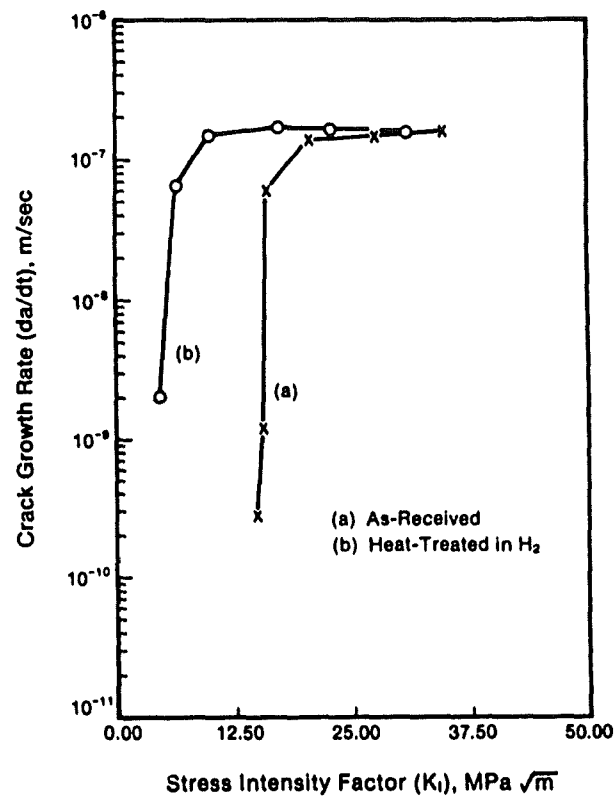
Intermittent immersion results indicate that the relative resistance to stress-corrosion cracking is the same as that found in weather exposure. The effect of the decreasing aluminum content on the resistance to stress-corrosion cracking is particularly pronounced in the weather exposure data (AZ61: 6Al-1Zn, AZ51: 5Al-1Zn, AZ31: 3Al-1Zn).

W-10 Tungsten Alloy: Effect of Heat Treatment on Stress-Corrosion Cracking Behavior, in Sodium Chloride



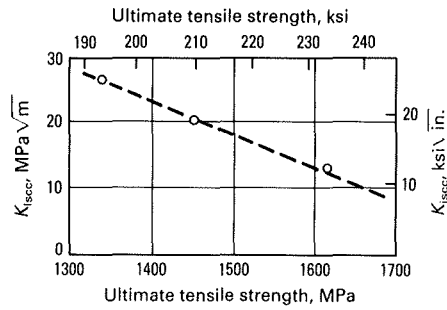
Effect of heat treatment on stress-corrosion cracking behavior in gaseous hydrogen of W-10 alloy in aerated 3% sodium chloride solution at open circuit potential.

W-10: 6.08Ni-2.87Fe bal W. The alloy is a duplex-phase alloy, consisting of a binder or matrix phase and tungsten grains, and is susceptible to stress-corrosion cracking. However, 97% W alloys tested under similar conditions were not susceptible to stress-corrosion cracking. At the open circuit potential, the matrix phase is anodic with respect to the tungsten grains. Stress-corrosion cracking of W-10 alloys occurs by anodic dissolution of the matrix phase. W-10 alloys are susceptible to hydrogen-assisted cracking when charged with hydrogen.



Influence of heat treatment in gaseous hydrogen on stress-corrosion cracking behavior of W-10 alloy in aerated 3% sodium chloride solution at $-1.2 V_{SCE}$.

Uranium Alloy U-0.75Ti: Plane-Strain Threshold for Stress-Corrosion Cracking Versus Ultimate Tensile Strength, in Cl⁻ Solution



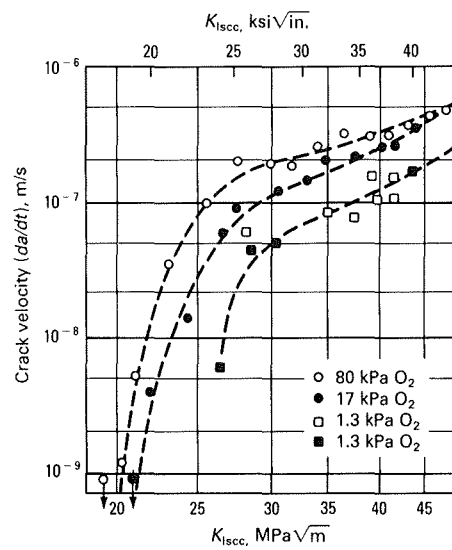
SCC thresholds for U-0.75Ti in air and in aqueous environments

Heat treatment	Yield strength		Environment	K_{Iscorr}	
	MPa	ksi		$MPa\sqrt{m}$	$ksi\sqrt{in.}$
380 °C (715 °F), 6 h	986	143	Dry air	42	38
380 °C (715 °F), 6 h	986	143	100% relative humidity air	27.5	25
380 °C (715 °F), 6 h	986	143	50 ppm Cl ⁻	17.6	16
380 °C (715 °F), 6 h	986	143	3.5% NaCl	17.6	16
Air cooled	607	88	Water	23	21
Air cooled	607	88	3.5% NaCl	16.5	15
450 °C (840 °F)	...	...	Air	28.6	26
450 °C (840 °F)	...	...	Water	15.4	14
450 °C (840 °F)	...	...	3.5% NaCl	11	10

Plane-strain threshold for stress-corrosion cracking versus ultimate tensile strength for U-0.75Ti in 50 ppm Cl⁻ solution at 25 °C.

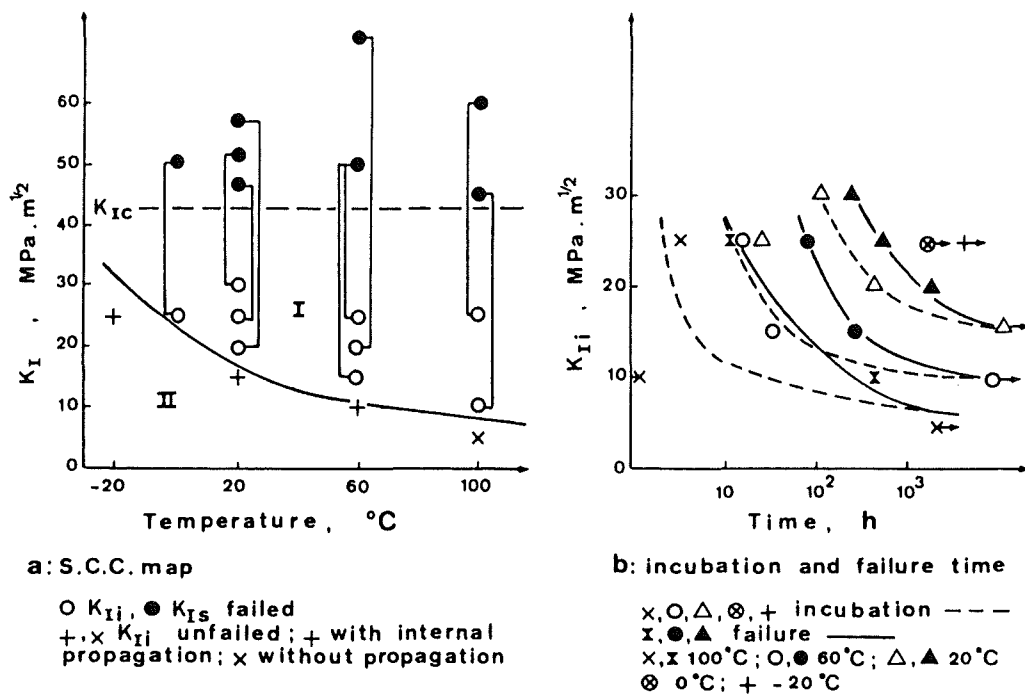
Uranium Alloy U-4.5Nb: Crack Velocity as a Function of Stress-Intensity Factor and Oxygen Pressure

525

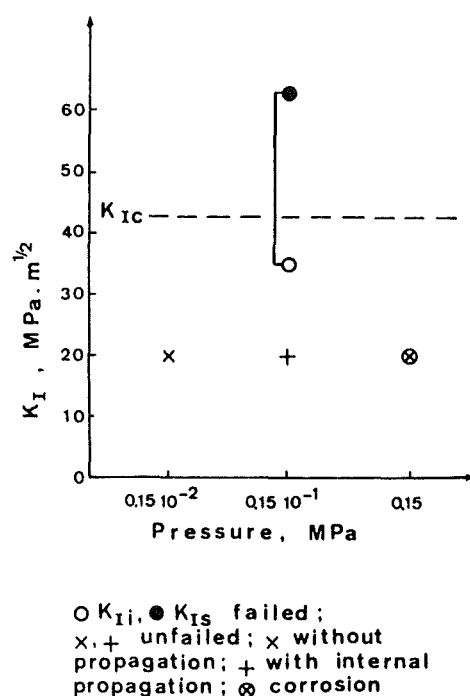


Crack velocity as a function of stress-intensity factor and oxygen pressure for U-4.5Nb.

Uranium Alloy U-7.5Nb-2.5Zr: Stress-Corrosion Cracking as a Function of Temperature, in Oxygen



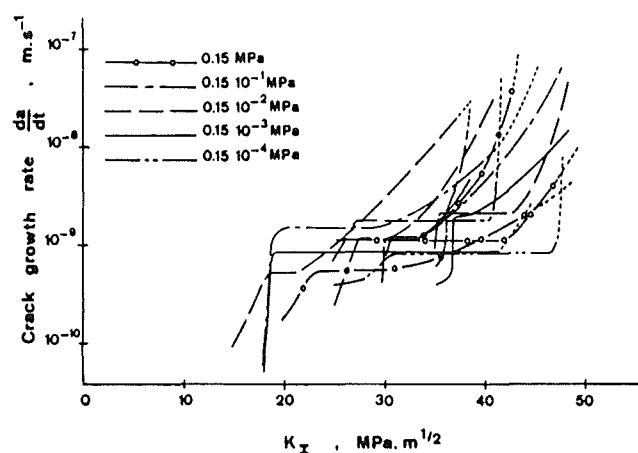
Stress-corrosion cracking tests of U-7.5Nb-2.5Zr in 0.15 MPa oxygen as a function of temperature.



Stress-corrosion cracking map for U-7.5Nb-2.5Zr tested in 0.15 MPa hydrogen as a function of temperature.

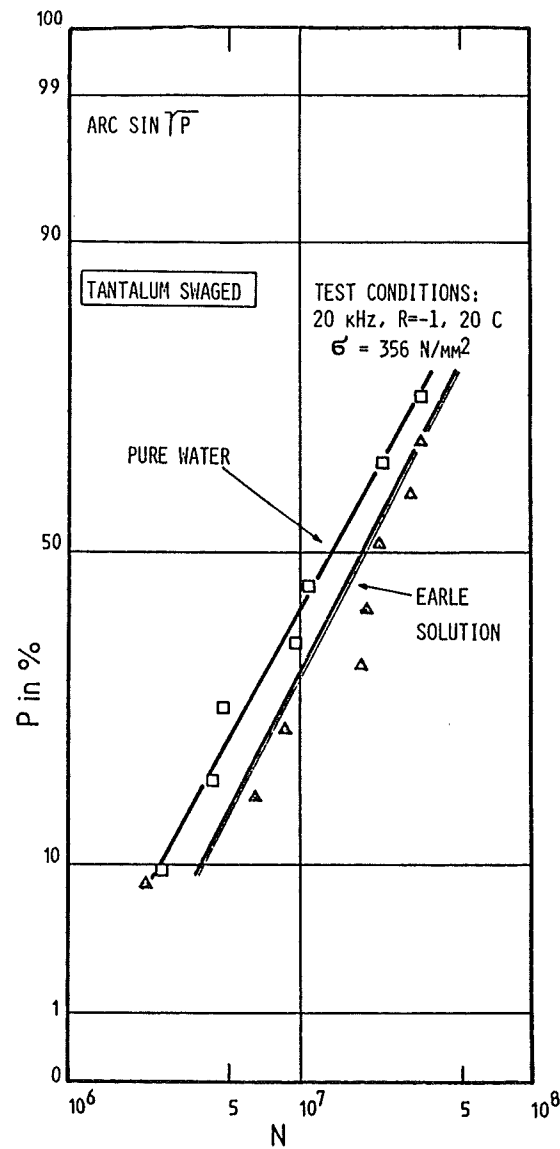
All specimens failed, even at no load. At -20 and +20 °C, corrosion occurs along the fatigue precrack, at the crack tip, and at some inclusions.

Uranium Alloy U-7.5Nb-2.5Zr: Crack Growth Velocity as a Function of Oxygen Pressure



Crack growth velocity in U-7.5Nb-2.5Zr as a function of oxygen pressure at 20 °C.

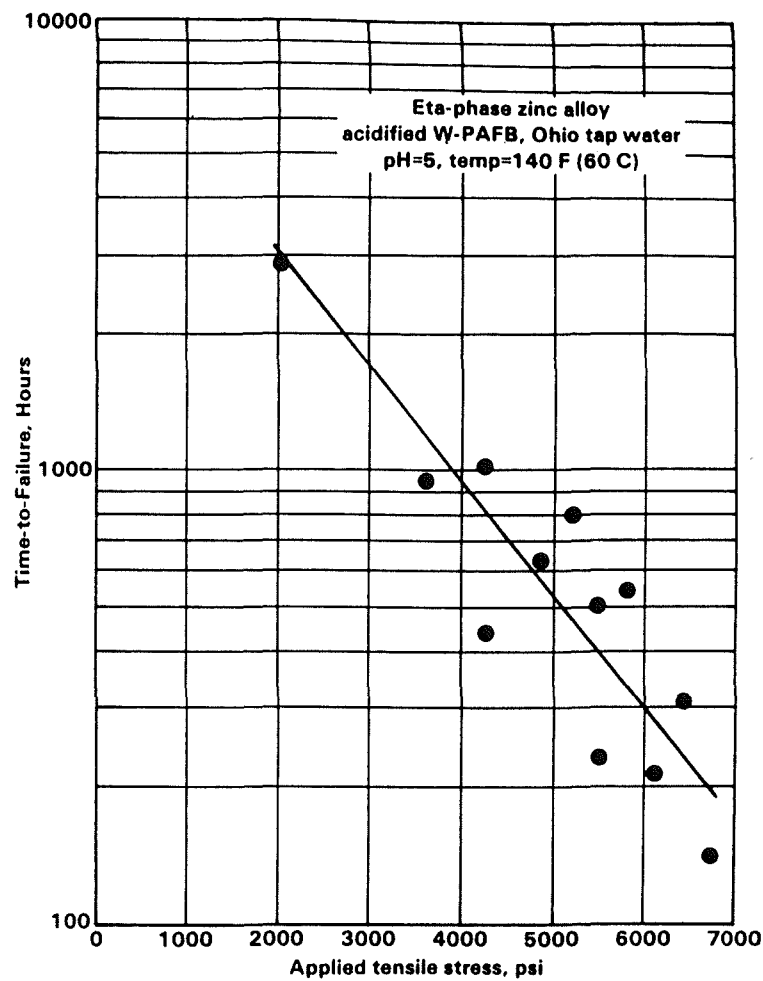
Source: D. Lepoutre, A.M. Nomine, and D. Miannay, Stress Corrosion Cracking Behavior in Gaseous Oxygen of a U-7.5Nb-2.5Zr Alloy, in Application of Fracture Mechanics to Materials and Structures, G.C. Sih, E. Sommer, and W. Dahl, Ed., Martinus Nijhoff, The Hague, 1984, p 907-918.



Effect of environment on fracture probability of tantalum specimens tested under corrosion-fatigue conditions.

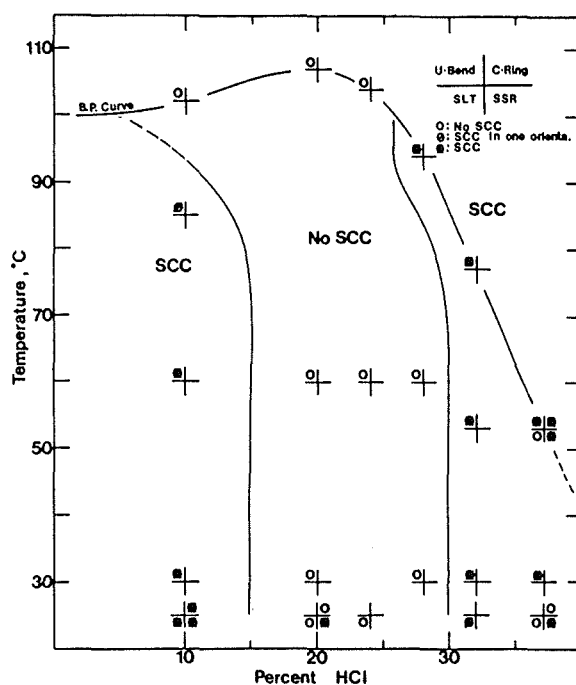
Environment: modified Earle's solution, 3% sodium chloride in water, pH 5 adjusted by additions of lactic acid.

Zinc: Time-to-Failure as a Function of Applied Stress Level, in Tap Water



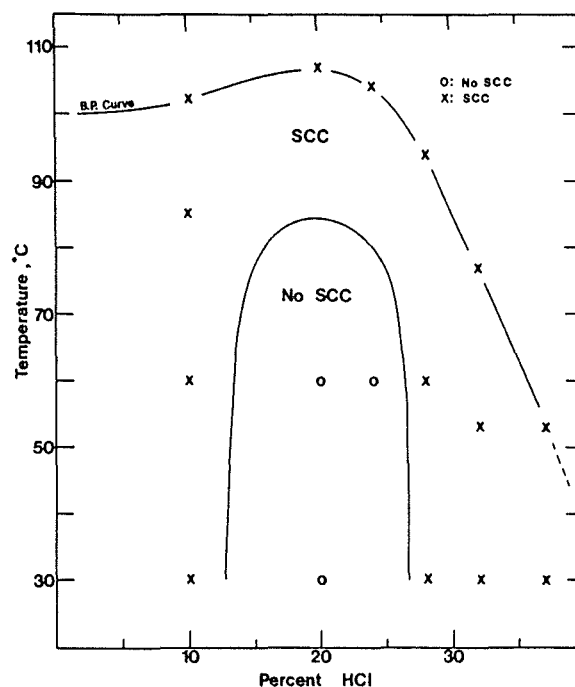
Time-to-failure of zinc at 60 °C in tap water as a function of applied stress level.

Source: C.J. Slunder and W.K. Boyd, Zinc: Its Corrosion Resistance, T.K. Christman and J. Payer, Ed., International Lead Zinc Research Organization, 1983, New York, p 153.

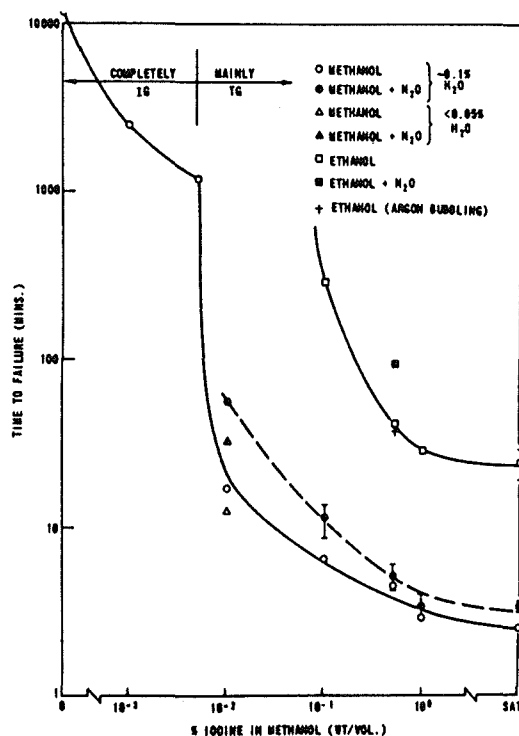


Effect of acid concentration and temperature on the stress-corrosion cracking of nonwelded specimens of zirconium in HCl solutions containing 500 ppm Fe^{3+} .

Zirconium: Effect of Acid Concentration and Temperature on Stress-Corrosion Cracking, in HCl Solutions Containing Fe^{3+}



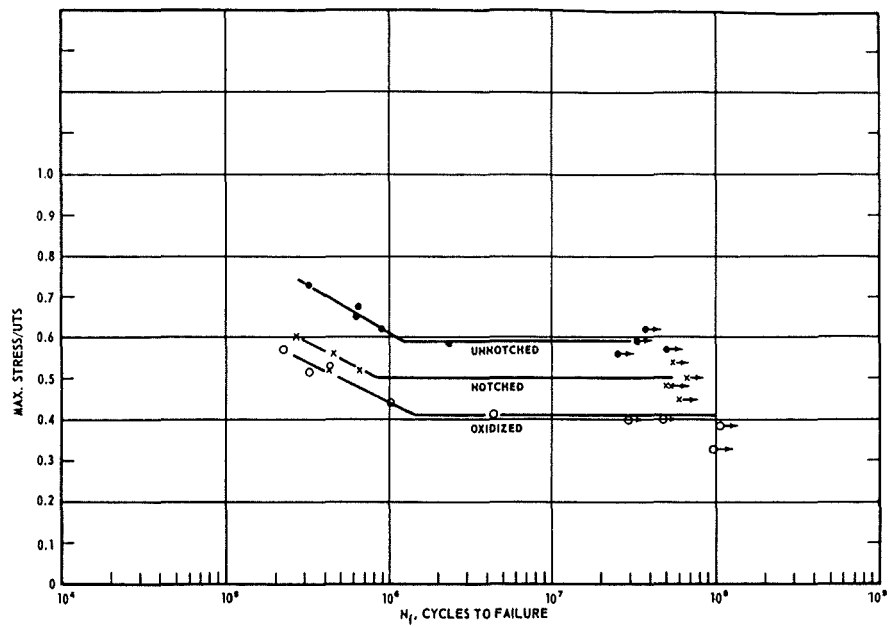
Effect of acid concentration and temperature on the stress-corrosion cracking of welded U-bends of zirconium in HCl solutions containing 500 ppm Fe^{3+} .



Effect of N_2O additions to alcoholic solutions of iodine on the time-to-failure of stressed split rings of Zircaloy.

A similar series of events takes place during the stress-corrosion cracking of zirconium alloys in halogen vapors, organic halogen solutions, and aqueous solutions of halogen and halide. In smooth specimens, a breakdown of the oxide film occurs, followed by intergranular chemical attack either by dissolution in solutions or by volatilization in halogen vapors. As the chemical attack penetrates, the stress-intensity rises until the critical stress-intensity for transgranular cracking is reached. Transgranular cracking is probably controlled by absorption of the halogen and not by chemical removal of the material or by hydrogen.

Zirconium Alloy Zircaloy-2: Fatigue Curves, in Ammoniated Water



Frequency 60 Hz $R = 0.1$

● Unnotched Specimens

○ Oxidized Unnotched Specimens

Ammoniated Water

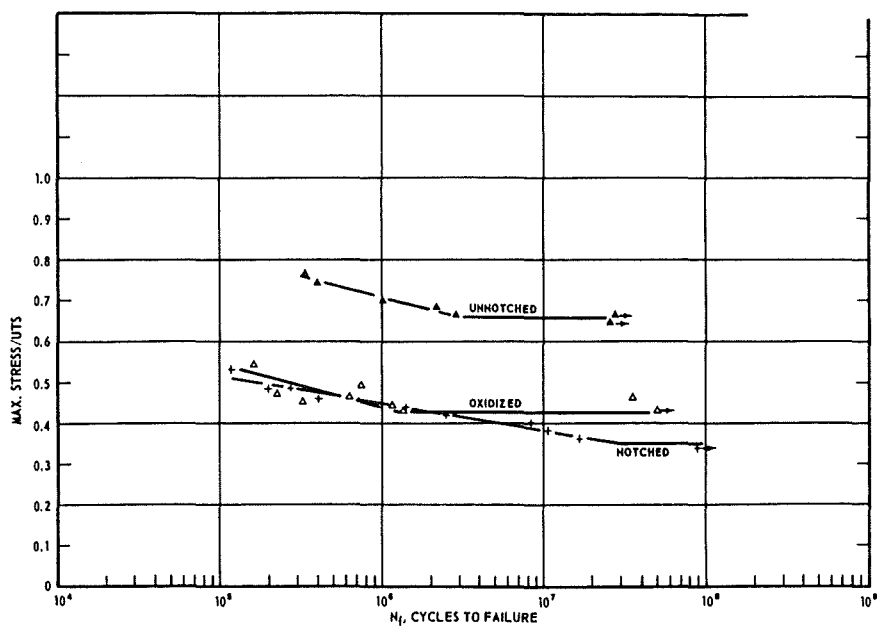
x Notched Specimens $K_t = 3$

→ No Failure

Fatigue curves for annealed Zircaloy-2 in ammoniated water at 573 K.

Zirconium Alloy Zircaloy-2: Fatigue Curves, in Lithiated Water

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Frequency 60 Hz $R = 0.1$

▲ Unnotched Specimens

△ Oxidized Unnotched Specimens

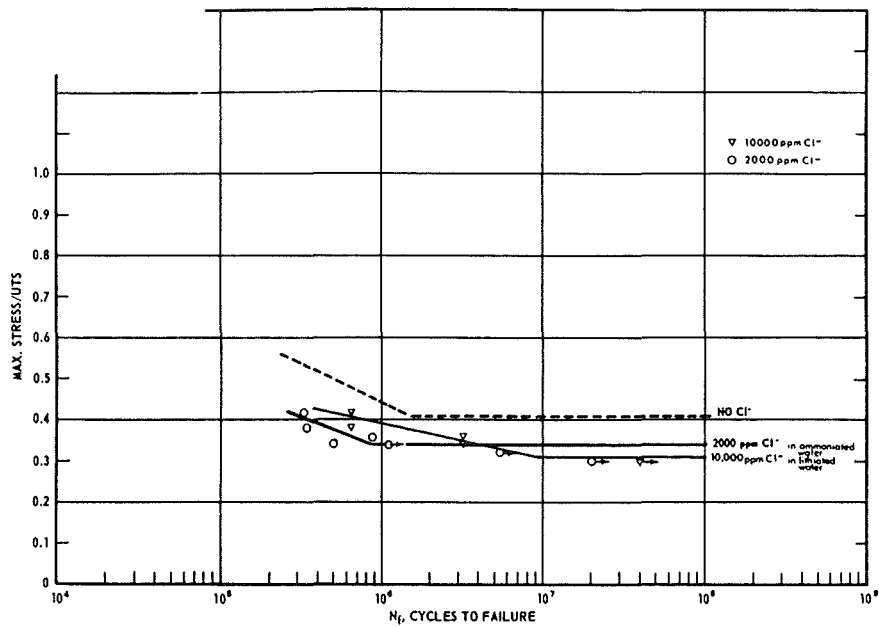
Lithiated Water

+ Notched Specimens $K_t = 3$

→ No Failure

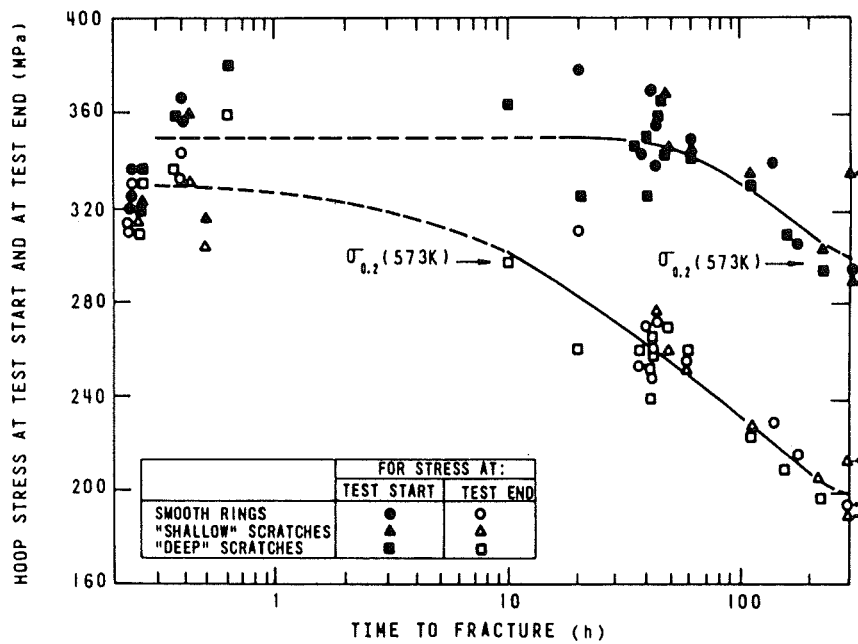
Fatigue curves for annealed Zircaloy-2 in lithiated water at 573 K.

Zirconium Alloy Zircaloy-2: Fatigue Curves, in Ammoniated Water and in Lithiated Water



Frequency 60 Hz R = 0.1 --- Ammoniated Water
 —○ Ammoniated Water + 2000 ppm Cl⁻
 —▽ Lithiated Water + 10,000 ppm Cl⁻
 → No Failure

Fatigue curves for oxidized unnotched Zircaloy-2 showing the effect of 2000 ppm Cl⁻ in ammoniated water and 10,000 ppm Cl⁻ in lithiated water at 573 K.

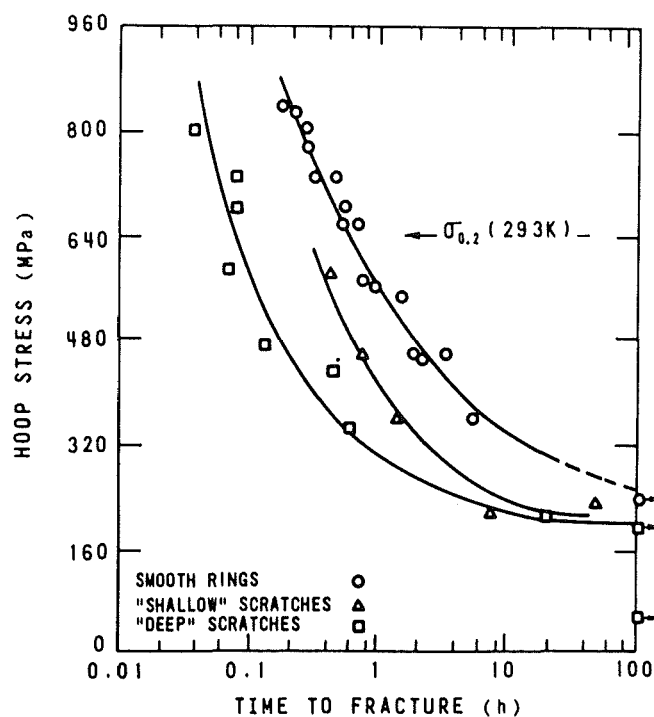


Time-to-failure for Zircaloy-4 rings (stress relieved at 683 K for 3 h) tested in an environment of iodine vapor (10 kg/m^3) at 573 K, some with scratched surfaces.

Zircaloy-2: 1.5% Sn, 0.12% Fe, 0.10% Cr, 0.05% N, bal Zr.

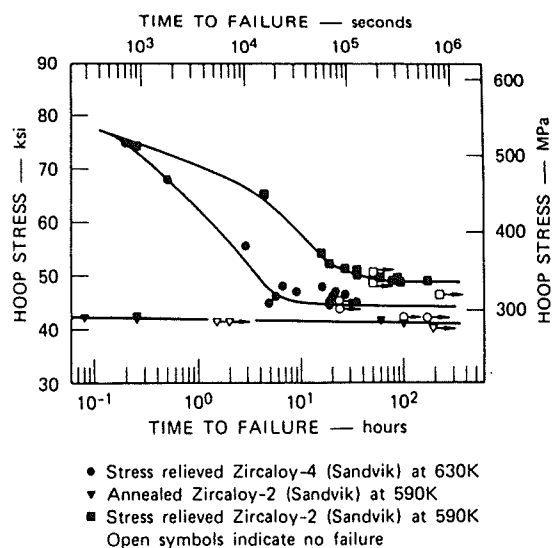
Zircaloy-4: 1.5% Sn, 0.12% O, 0.21% Fe, 0.10% Cr, bal Zr.

Zirconium Alloy Zircaloy-4: Time-to-Failure, in Methanol + Iodine Solution



Effect of scratching the inside surface of stressed slotted rings of stress relieved Zircaloy-4 tubing on time-to-failure in a solution of methanol (1 liter) and iodine (50 g) at 293 K.

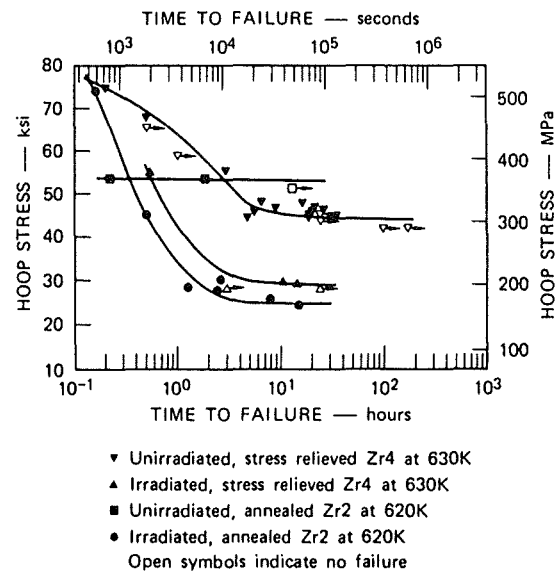
Source: R. Sejnoha and J.C. Wood, Iodine-Induced Stress Corrosion Cracking of Fixed Deflection Stressed Slotted Rings of Zircaloy Fuel Cladding, in Zirconium in the Nuclear Industry, STP 681, American Society for Testing and Materials, Philadelphia, 1979, p 261-284.



Results of iodine stress-corrosion cracking tests at ~600 K on internally pressurized tube specimens of unirradiated Zircaloy-2 and Zircaloy-4.

Zircaloy-2: 1.5% Sn, 0.12% Fe, 0.10% Cr, 0.05% N, bal Zr.
 Zircaloy-4: 1.5% Sn, 0.12% O, 0.21% Fe, 0.10% Cr, bal Zr.

Zirconium Alloys Zircaloy-2 and Zircaloy-4: Effect of Irradiation

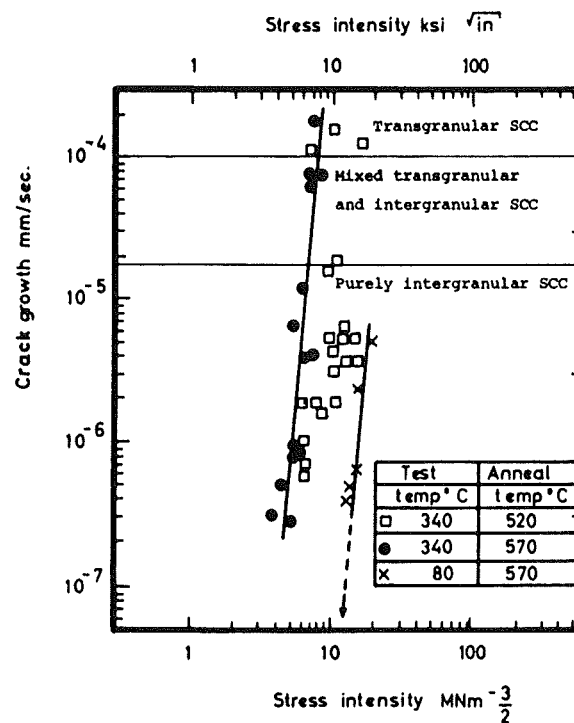


**Effect of irradiation on the resistance of internally pressurized
Zircaloy tube to iodine-induced failure at ~600 K.**

Source: R. Sejnoha and J.C. Wood, Iodine-Induced Stress Corrosion Cracking of Fixed Deflection Stressed Slotted Rings of Zircaloy Fuel Cladding, in Zirconium in the Nuclear Industry, STP 681, American Society for Testing and Materials, Philadelphia, 1979, p 261-284.

Zircaloy Cladding Tube: Stress-Corrosion Cracking Velocity as a Function of Stress-Intensity

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Stress-corrosion cracking velocity as a function of the stress-intensity factor for fatigue precracked Zircaloy cladding tube (3 mg iodine per cm³).

INTRODUCTION

INTRODUCTION

Stress-corrosion cracking (SCC) designates failure by cracking under the combined action of corrosion and an applied or a residual stress. The crack path may be either intergranular or transgranular, depending on the metal and the corrosive medium.

Corrosion fatigue (CF) is the result of the application of repeated or fluctuating stresses in a corrosive environment and is usually characterized by a shorter life than would be encountered as a result of repeated or fluctuating stresses alone or the corrosive environment alone.

Environmentally assisted cracking (EAC) is a generic term that encompasses stress-corrosion cracking and corrosion fatigue, as well as liquid metal embrittlement (LME) and hydrogen-assisted cracking (HAC). The environment may be solid, liquid, or gaseous.

In this introductory section, the principal features of stress-corrosion cracking and corrosion fatigue will be presented, together with a brief review of the various test methods used to study these time-dependent processes and a summary of the mechanisms involved. In general, the means of applying stress, whether steady or cyclic, are relatively straightforward, as are the purely mechanical consequences of the application of these stresses in terms of plastic deformation and crack tip behavior, although even in these areas we continue to learn more of the mechanisms involved. The real complexities arise when consideration is given to the manner in which the environment interacts with the mechanical processes to initiate and propagate the cracks that result in failure. Whereas a background in mechanics and metallurgy is sufficient to deal with the mechanical aspects of fatigue, additional background in chemical thermodynamics, corrosion, and electrochemistry is needed to deal comprehensively with the broad field of environmentally assisted cracking. However, even though there are difficulties in understanding the various aspects of environmentally assisted cracking in detail, largely because the small scale of critical reactions prevents direct observation, nevertheless considerable progress has been made in recent years in understanding the nature of environmentally assisted cracking both in the case of stress-corrosion cracking and in corrosion fatigue, as is demonstrated by the information presented in this Atlas.

ASM International has included a comprehensive treatment of environmentally assisted cracking in Volume 13 of the Ninth Edition of the *Metals Handbook* entitled "*Corrosion*" (Ref 1), which contains much useful additional background information. This Atlas differs from Volume 13 of the *Metals Handbook* in that a much more extensive collection of graphs from the literature on the specific topics of stress-corrosion cracking and corrosion fatigue is presented. It is hoped that this compilation of data will be useful to engineers and researchers in need of information concerning these matters. In this Atlas, the stress-corrosion cracking and corrosion fatigue graphs are grouped by material classifications, such as "Copper-Base Alloys." There is an introduction to each of these sections, which describes aspects of stress-corrosion cracking and corrosion fatigue that are pertinent to that particular group of alloys. It is also noted that the *Atlas of Fatigue Curves* (Ref 2), also published by ASM International, serves as a useful companion to the present work. In addition to an extensive compilation of fatigue curves, it also contains information on fatigue test procedures and specimen design.

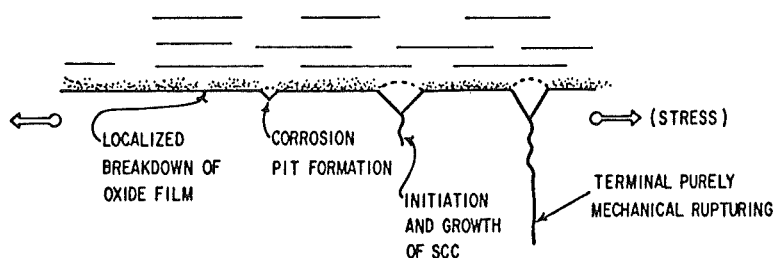


Fig. 1 Sequence of events (left to right) in stress-corrosion of an initially smooth specimen (Ref 3).

As indicated above, stress-corrosion cracking requires the presence of an aggressive environment. However, such an environment may be as seemingly innocuous as distilled water. Similarly, corrosion fatigue also involves an aggressive environment, and in this case, laboratory air may suffice. In fact, the

vast majority of fatigue tests are actually corrosion fatigue tests, with a component of ambient air—usually water vapor and oxygen—being involved in surface reactions that reduce the fatigue life. Only in a vacuum or in a completely dry inert gas environment can the corrosive effects of the atmosphere be avoided. However, there are times when reactions with the environment can be beneficial rather than deleterious. For example, the formation of a protective oxide at an elevated temperature may improve resistance to crack initiation and growth. Even at room temperature, oxides form on fracture surfaces and increase the crack opening level, with an attendant reduction in the rate of crack growth in the near-threshold region.

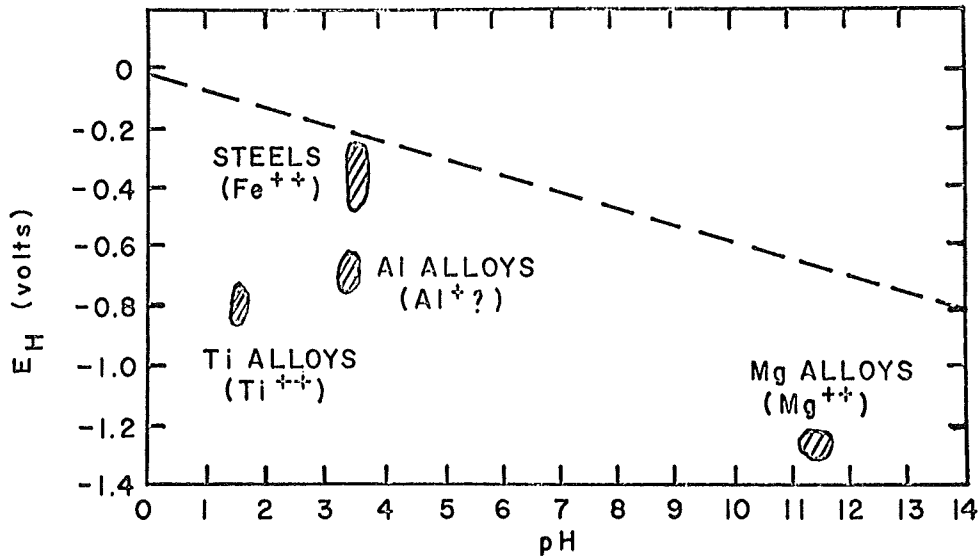


Fig. 2 Potential and pH data for the tips of growing stress-corrosion cracks of four alloy systems in nearly neutral salt water. The dashed line indicates the conditions below which hydrogen can be reduced from water (Ref 4).

The way in which stress and the environment work together to bring about cracking is indicated in Fig. 1, which shows a generalized sequence of the events that take place during the course of a stress-corrosion cracking process (Ref 3). It is now well established that when an aqueous environment is involved, the chemical makeup of the solution at the crack tip, including its pH, may differ significantly from that of the bulk solution. For example, Fig. 2 shows the crack tip pH values for four alloy systems plotted as a function of the most active potential exhibited by the alloys in nearly neutral (pH 7) 3.5% sodium chloride solution at room temperature (Ref 4). Also shown in Fig. 2 is a line below which it is thermodynamically possible for hydrogen to evolve from water. For all four alloy systems, when the alloy is undergoing stress-corrosion cracking, the conditions of potential and pH are favorable for the cathodic reaction to be the reduction of hydrogen. Also included in Fig. 2 is the stable soluble species of the principal element of the alloy at the potential-pH conditions of the crack tip. In the case of aluminum, the soluble species is probably Al^{+++} .

Figure 3 shows how the crack tip pH values for high-strength steels vary with applied potential in neutral 3.5% sodium chloride solution (Ref 4). In all cases, the conditions at the crack tip are favorable for the hydrogen reduction process, although based only upon the pH value of the bulk solution this may not appear to be the case. The properties of the bulk solution, on the other hand, may still be important in the crack initiation phase. In addition, the presence of corrosion products on the crack faces may also influence the rate of crack propagation, in some cases by actually wedging open the crack, and in others by interfering with circulation between the bulk environment and the crack tip region. In corrosion fatigue, however, differences between the chemistry of the crack tip region and the bulk solution may not occur to the same degree, because a pumping action develops during the opening and closing of the crack, which tends to keep the crack tip chemistry similar to that of the bulk. This mixing becomes more effective with higher cyclic frequency.

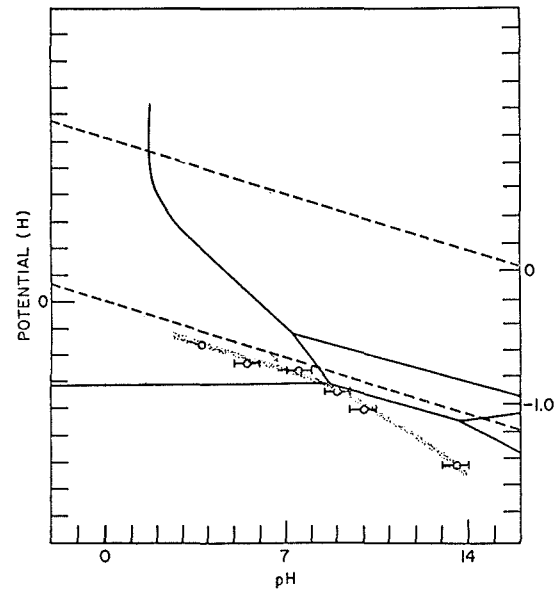


Fig. 3 pH at the tips of growing stress-corrosion cracks a low-alloy steel tested in neutral 3.5% sodium chloride solution; specimens potentiostatted to various levels (Ref 4).

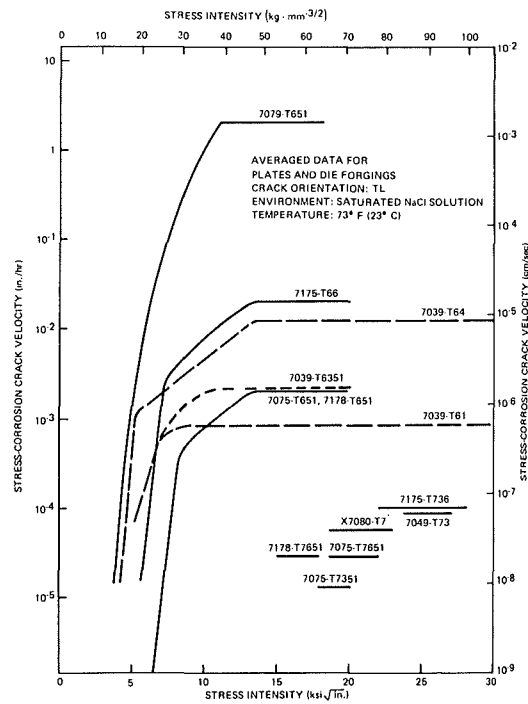


Fig. 4 Effect of stress-intensity on stress-corrosion crack-velocity for several high-strength aluminum alloys immersed in saturated aqueous sodium chloride solutions (Ref 5).

Although the results of testing in the laboratory are useful in identifying the circumstances that can lead to stress-corrosion cracking or corrosion fatigue, and also in making comparisons between materials, as indicated in Fig. 4, it is usually not a straightforward matter to apply information gained in the laboratory directly to the prediction of service lifetimes (Ref 5). One reason for this is that many laboratory tests are intended more to screen for stress-corrosion cracking or corrosion fatigue susceptibility in a relatively short time rather than to provide a basis for lifetime estimation in service. In addition, the loading and environmental conditions experienced in service are often not well defined. One exception is the study of the behavior of steels used in nuclear pressure vessels where service conditions are carefully simulated in the laboratory.

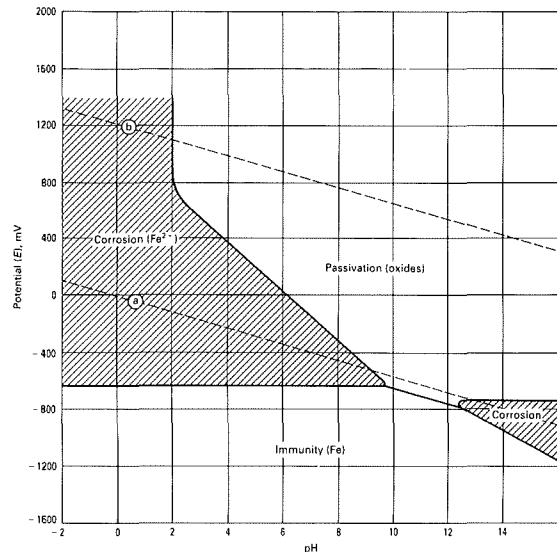


Fig. 5 Potential-pH (Pourbaix) diagram for iron in water.

One of the guidelines for setting allowable stresses in the case of stress-corrosion cracking is the threshold level for crack propagation, often designated as $K_{I_{SCC}}$, where K_I is the stress-intensity factor for Mode I loading at threshold for a particular combination of alloy and environment. A similar parameter, $\Delta K_{I_{th}}$, is used in the case of corrosion fatigue, where ΔK is the range of the stress-intensity factor. Unfortunately from the designer's point of view, these thresholds only apply to cracks of macroscopic size. Cracks of microscopic dimensions can nucleate at stress-intensity levels below these thresholds, because of the inapplicability of linear elastic fracture mechanics (LEFM) to the treatment of cracks of the size of 1 ml or less. The threshold stress-intensity value at threshold is related to a product of stress and the square root of the crack length. This product is a constant for macroscopic cracks. However, if we attempt to apply this to the case of small cracks, as the crack length approaches zero the stress must rise to unrealistically high values. In reality, there is a transition with increasing crack length from the threshold stress for a smooth specimen, a much lower stress than expected on the basis of linear elastic fracture mechanics analysis, to the condition for the threshold of a macroscopic crack. To further complicate the analysis of crack growth behavior in the early stages of crack growth in stress-corrosion cracking and corrosion fatigue, there may be a multiplicity of cracks, some of which grow and join, and some of which may even become dormant. Furthermore, additional cracks may nucleate at various times during exposure. In view of the uncertainties involved in crack formation and growth processes wherein the environment is a factor, it is clear that past experience is a useful asset in predicting future behavior.

After having pointed out some of the pitfalls of lifetime predictions, we next consider the laboratory procedures used in the characterization of stress-corrosion cracking and corrosion fatigue susceptibility. One of the basic aids is a potential-pH (Pourbaix) diagram. An example of a Pourbaix diagram is shown in Fig. 5, in this case for the water-iron system. The region between the dashed lines of Fig. 5 is a region wherein water is stable. Above the dashed line "b," oxygen is evolved, and below the dashed line "a," hydrogen is evolved. A Pourbaix diagram represents conditions of thermodynamic equilibrium and as such does not deal directly with the kinetics of reactions or with the presence of metastable phases. The diagrams do indicate which processes can occur and how strong the tendency is for changes to take place.

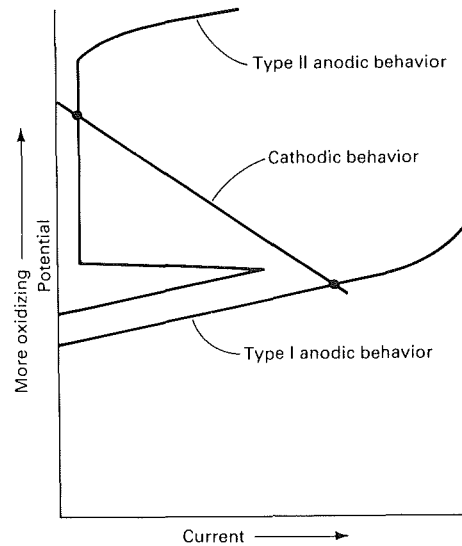


Fig. 6 Schematic showing the two basic types of electro-chemical response exhibited by nickel-base alloys (Ref 6).

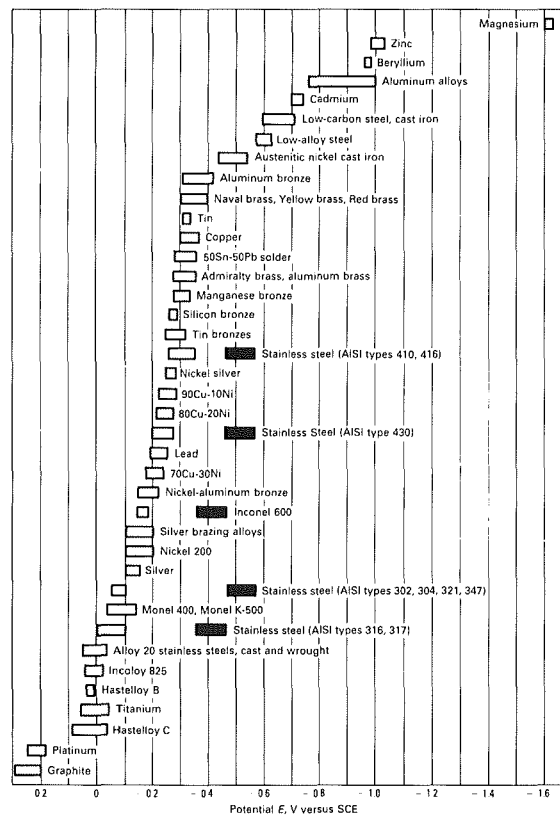
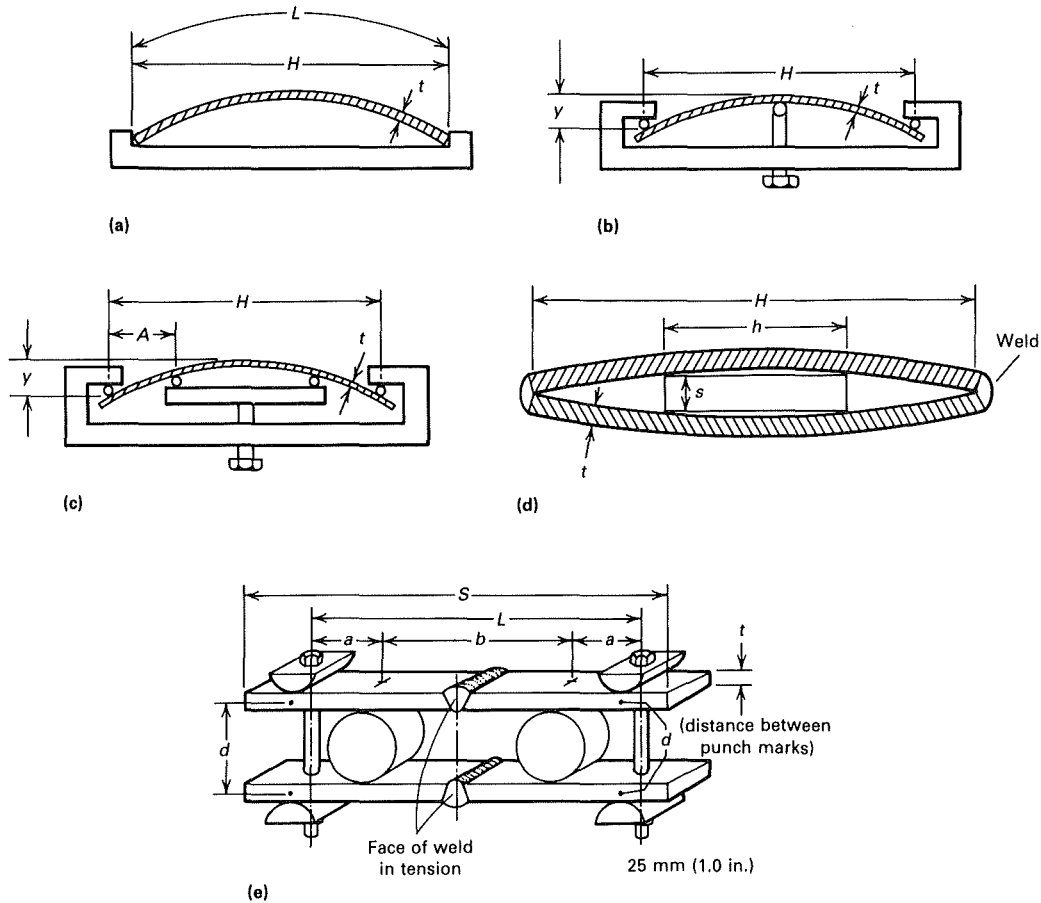


Fig. 7 Corrosion potentials of various metals and alloys in flowing seawater at 10 to 25 °C. Those metals that exhibit a black bar may become active and exhibit a potential near $-0.5 V_{SCE}$ in low-velocity or poorly aerated water and in shielded areas (Ref 7 and 8).

Another basic aid is the potential-current (Evans) diagram. Figure 6 shows a schematic Evans diagram in which two types of behavior for metals in solution are indicated. In Type I behavior, the corrosion or anodic current increases monotonically with potential. In Type II behavior, the anodic current initially increases with potential (active behavior), then decreases to a small constant value (passive behavior), and finally increases again (transpassive behavior). These two types of behavior are not intrinsic properties of an alloy, but are the result of interaction of the alloy with the environment. For example, a nickel-molybdenum alloy exhibits Type I behavior in dilute H_2SO_4 and HCl and exhibits Type II behavior in concentrated H_2SO_4 (Ref 6).



Bolt loaded double-beam specimen dimensions for various plate thicknesses

t		a		b		L		S	
mm	in.	mm	in.	mm	in.	mm	in.	mm	in.
3.2	0.125	100	4.0	50	2.0	250	10.0	305	12.0
6.4	0.25	100	4.0	50	2.0	250	10.0	305	12.0
9.5	0.375	120	4.75	90	3.5	330	13.0	380	15.0
13	0.5	120	4.75	90	3.5	330	13.0	380	15.0
19	0.75	140	5.5	150	6.0	430	17.0	480	19.0
25	1.0	150	6.0	200	8.0	510	20.0	560	22.0
38	1.5	165	6.5	305	12.0	635	25.0	685	27.0

Fig. 8 Schematic specimen and holder configurations for bent-beam stressing. (a) Two-point loaded specimen. (b) Three-point loaded specimen. (c) Four-point loaded specimen. (d) Welded double-beam specimen. (e) Bolt-loaded double beam specimen (Ref 10).

The information from a galvanic series in the appropriate medium is also useful when the only information needed is which of the materials in a system are possible candidates for galvanically accelerated stress-corrosion cracking or corrosion fatigue, and which will be unaffected or protected. Figure 7 gives the corrosion potentials of various metals and alloys in flowing seawater at 10 to 25 °C (50 to 80 °F) at a flow rate of 2.5 to 4 m/s. The alloys are listed in order of potential versus the saturated calomel reference electrode (SCE) that they exhibited. The zero potential of the reference electrode is -0.268 V below the zero potential of the standard hydrogen electrode (SHE). Those metals and alloys indicated by black bars may become active and exhibit a potential near -0.5 V_{SCE} in low-velocity or poorly aerated water and in shielded areas (Ref 7 and 8). Unfortunately, galvanic series have been established with such completeness in very few environments.

Tests and Results

Most specimen configurations used for stress-corrosion cracking testing fall into the following three categories (Ref 9): smooth, statically loaded specimens; notched or precracked specimens; statically loaded specimens; and smooth, dynamically loaded specimens. Corrosion fatigue specimens include both smooth and notched configurations. The types of results obtained with each of these different categories of specimens will next be considered.

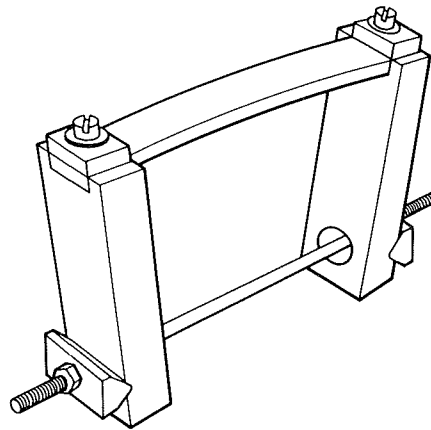


Fig. 9 Bent beam designed to produce pure bending (Ref 10).

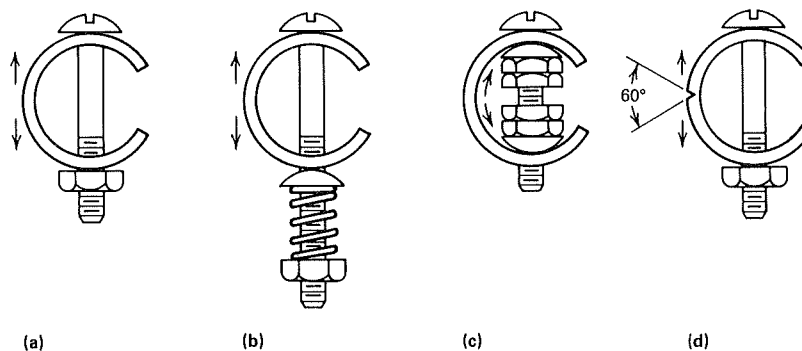


Fig. 10 Methods of stressing C-rings. (a) Constant strain. (b) Constant load. (c) Constant strain. (d) Notched C-ring; a similar notch could be used on the side of (a), (b), or (c) (Ref 10).

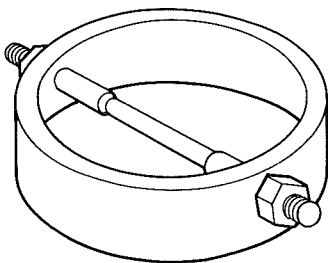
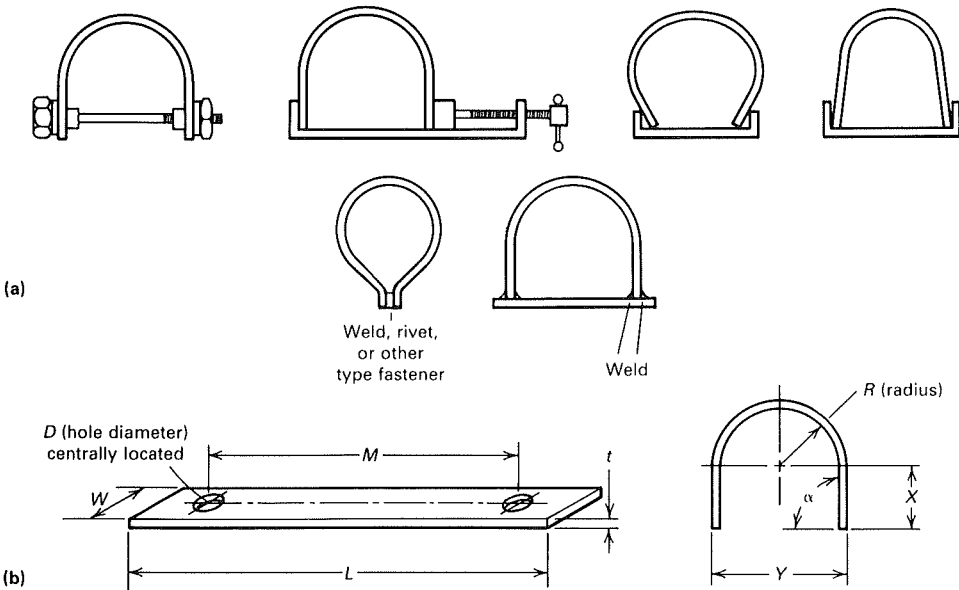


Fig. 11 Ring-stressed specimen for field testing (Ref 10).



Alternative size	L		M		W		t	
	mm	in.	mm	in.	mm	in.	mm	in.
A	80	3.2	50	2.0	20	0.8	2.5	0.098
B	100	4.0	90	3.5	9	0.35	3.0	0.12
C	120	4.7	90	3.5	20	0.8	1.5	0.06
D	130	5.1	100	4.0	15	0.6	3.0	0.12
E	150	5.9	140	5.5	15	0.6	0.8	0.03
F	310	12.2	250	9.8	25	0.98	13.0	0.51
G	510	20.1	460	18.1	25	0.98	6.5	0.26

Alternative size	D		X		Y		R	
	mm	in.	mm	in.	mm	in.	mm	in.
A	10	0.4	32	1.26	14	0.55	5	0.2
B	7	0.28	25	0.98	38	1.50	16	0.6
C	8	0.31	35	1.4	35	1.4	16	0.6
D	6	0.24	45	1.77	32	1.26	13	0.51
E	3	0.12	61	2.40	20	0.8	9	0.35
F	13	0.51	105	4.13	90	3.5	32	1.26
G	13	0.51	136	5.35	165	6.5	76	3.0

Note: $\alpha = 1.57$ rad

Fig. 12 Typical U-bend stress-corrosion cracking specimens. (a) Various methods of stressing U-bends. (b) Typical U-bend specimen dimensions (Ref 10).

Category I: Smooth, Statically Loaded Specimens

This category includes U-bends, C-rings, bent beams, and dead load tensile bars (ASTM G30, G35, G36, G38, and G47). Figures 8 through 13 (Ref 10) show a variety of test specimen configurations that may be used under constant load or constant displacement conditions. During a constant displacement test, the stress levels relax as cracks form, whereas in constant load tests the reverse is true. In stress-corrosion cracking testing of smooth specimens, the time-to-failure as a function of applied stress level or displacement level is obtained, with results plotted as in Fig. 14 for different initial stresses. For specimens as in Fig. 12, the test may simply be a quality control test to determine if the specimen will endure a predetermined number of hours in a particular environment.

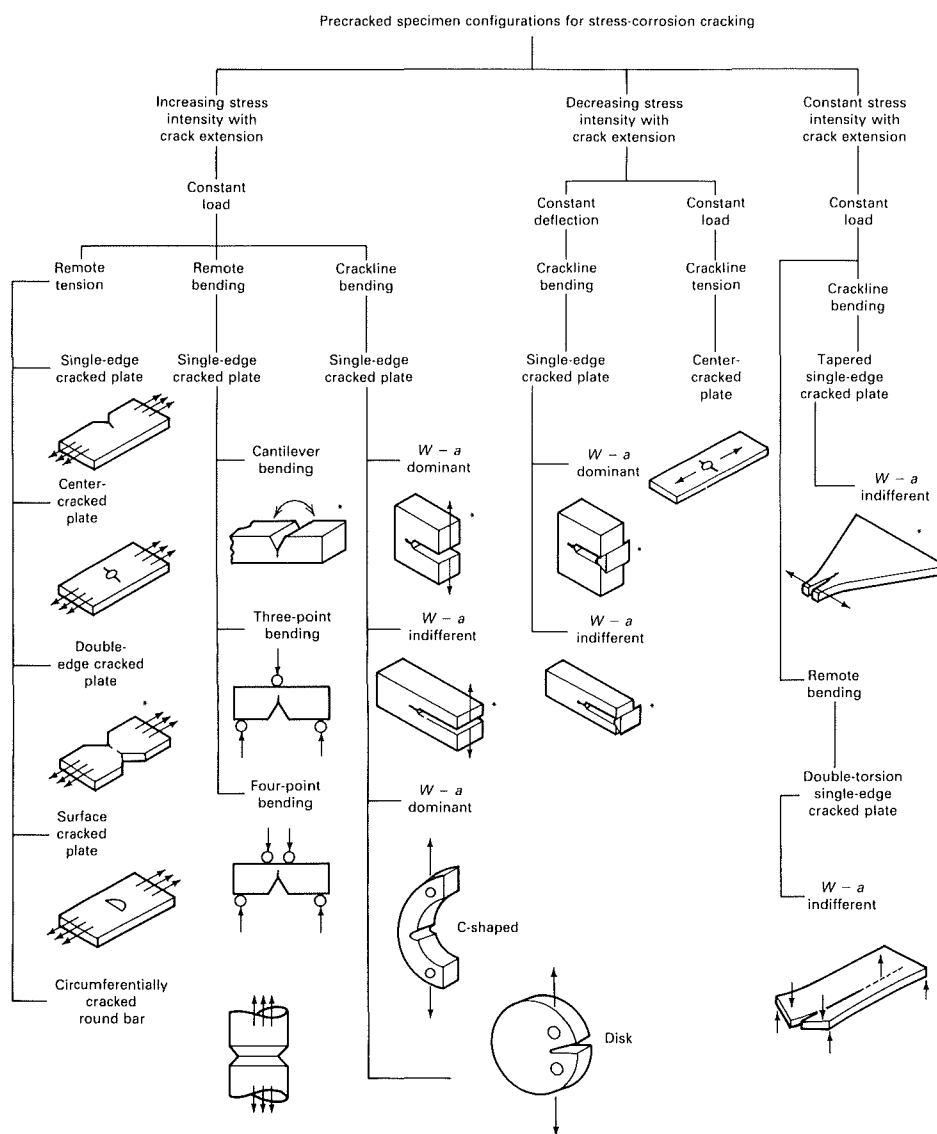


Fig. 13 Classification of precracked specimens for stress-corrosion cracking testing. Asterisks denote commonly used test configurations (Ref 10).

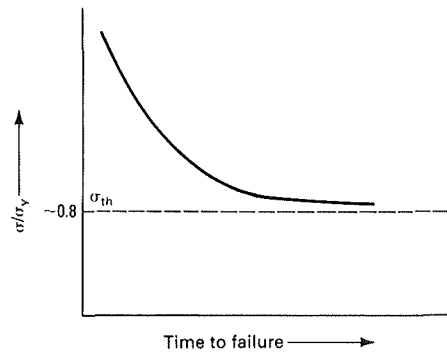


Fig. 14 Schematic diagram of time-to-failure versus applied stress, σ , normalized to the yield strength, σ_y , for stress-corrosion cracking.

In corrosion fatigue, the usual practice is to test under constant load amplitude conditions at a number of load levels to obtain results such as shown in Fig. 15. The loading conditions for fatigue testing are more varied than for stress-corrosion cracking testing. For example, the mean stress is a variable, and the designations "R" (equal to the ratio of the minimum to maximum stress in a cycle) and "A" (equal to the ratio of the stress amplitude to the mean stress) are used to identify the nature of the loading cycle. R and A are interrelated through the equation:

$$R = (1 - A) / (1 + A) \quad (1)$$

In addition, there is more attention given to the effects of variable amplitude loading in fatigue than is the case with stress-corrosion cracking, although in service applications involving the potential for stress-corrosion cracking the load levels can vary. In estimating lifetimes under variable amplitude loading in either stress-corrosion cracking or corrosion fatigue, in first approximation it can be assumed that linear damage rules apply. That means for corrosion fatigue at a given frequency failure will occur when the sum of the fractional cyclic lifetimes spent at each amplitude is equal to unity. A similar rule applies for stress-corrosion cracking.

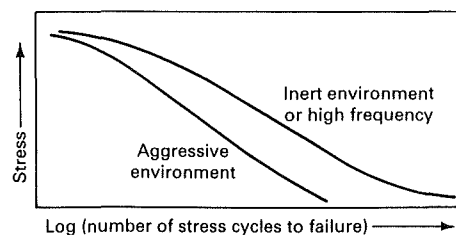


Fig. 15 Comparison of S/N curves for a material in an inert environment and in an aggressive environment.

Category 2: Notched and Precracked Specimens

This category includes specimens such as cantilever beams, compact tension specimens, and double cantilever beams (ASTM E399). Figure 13 shows a variety of such specimens, which are used in stress-corrosion cracking studies as well as corrosion fatigue tests. A schematic of a wedge-opening load specimen used in stress-corrosion cracking testing is shown in Fig. 16. A labeling system to indicate the orientation of specimens in the parent product is shown in Fig. 17. Because the stress-corrosion cracking properties are often lowest in the short-transverse direction, loads are often applied in that direction to obtain the most conservative results.

If notched specimens are used to study stress-corrosion cracking or hydrogen embrittlement, results as in Fig. 18 (Ref 11) are obtained.

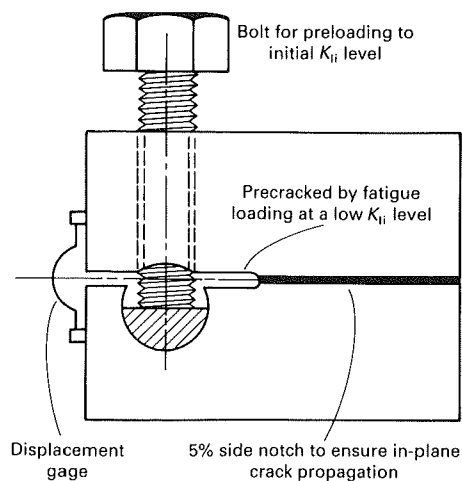


Fig. 16 Schematic showing the basic principle of a wedge-opening load specimen.

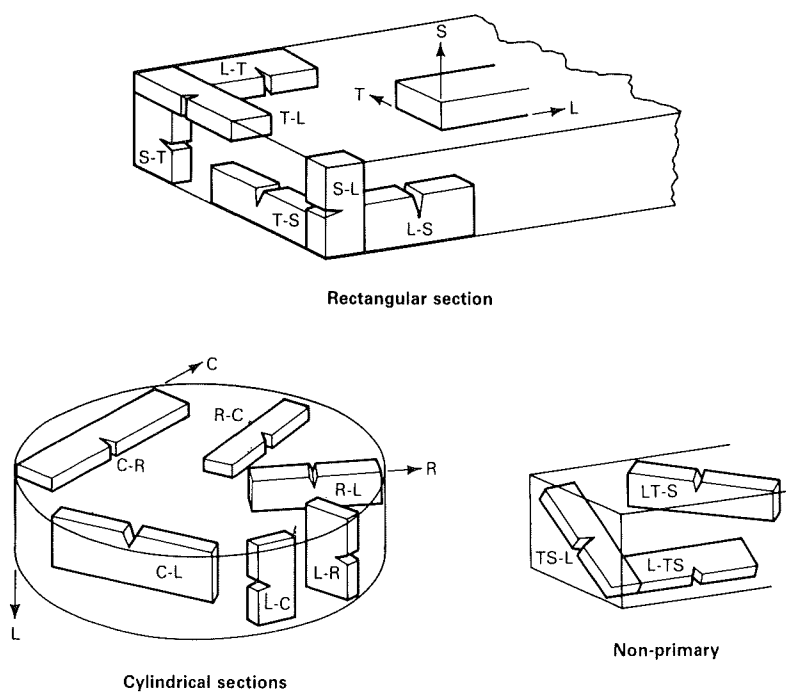


Fig. 17 Specimen orientation and fracture plane identification. L, length, longitudinal, principal direction of metalworking (rolling, extrusion, axis of forging); T, width, long-transverse grain direction; S, thickness, short-transverse grain direction; C, chord of cylindrical cross section; R, radius of cylindrical cross section. First letter: normal to the fracture plane (loading direction); second letter: direction of crack propagation in the fracture plane (Ref 10).

In studying crack growth characteristics in the linear elastic range, i.e., where the scale of the plastically deformed region at the crack tip is small with respect to other specimen dimensions as specified by ASTM E399, it is customary to use the stress-intensity factor, K , as a correlating parameter. K is defined as:

$$K = \lim_{\rho \rightarrow 0} K_t \sigma \sqrt{\frac{\pi \rho}{4}} \quad (2)$$

where K_t is the theoretical stress concentration factor, σ is the applied stress, and ρ is the radius of the tip of the stress raiser.

The local elastic stress field at a crack tip is related to K through equations of the type:

$$\sigma_{yy} = \sqrt{\frac{K}{2\pi r}} \xi(\theta) \quad (3)$$

where σ_{yy} is the local stress in the loading direction, and r and θ are polar coordinates measured from the crack tip.

The value of K is a function of the applied load or applied displacement, the crack length, and the specimen geometry. As an example, the expression for K for the commonly used compact tension specimen shown at the top of the third column in Fig. 13 is:

$$K = P/BW^{1/2} \cdot f(a/W) \quad (4)$$

$$f(a/W) = [(2 + a/W)(0.886 + 4.64 a/W) - 13.32 a^2/W^2 + 14.762 a^3/W^3 - 5.6 a^4/W^4]/(1 - a/W)^{3/2}$$

where P is the applied load, B is the specimen thickness, W is the specimen width measured from the load line, and a is the crack length, also measured from the load line.

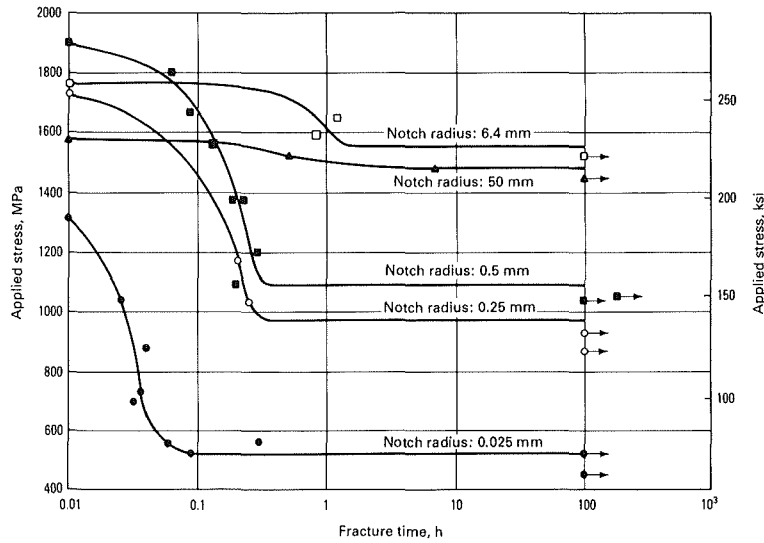


Fig. 18 Static fatigue curves for steel embrittled by hydrogen showing the effect of notch sharpness (Ref 11).

The nature of test results will depend on the nature of the loading method used. In Fig. 19(a), note that, as a stress-corrosion crack grows in a cantilever specimen under constant load, the K level rises throughout the test, and the test terminates when the level of the fracture toughness is reached. In contrast, under constant displacement loading, the K level decreases during the test until the threshold level for propagation, K_{Isc} , is reached (Fig. 19b). In a decreasing K test, it is frequently observed, as shown in Fig. 19(b), that the initial crack velocity may be constant. Figure 20 is an overall schematic that illustrates the relationship between the applied K level and the velocity of crack growth. In this figure, there are two regions in which the crack growth rate is K dependent. These regions are separated by a plateau region

in which the velocity is either constant or almost independent of K . In the plateau region, the lack of dependency of the crack velocity on K occurs because the rate of crack growth is limited by the rate of diffusional processes taking place in the solution at the crack tip.

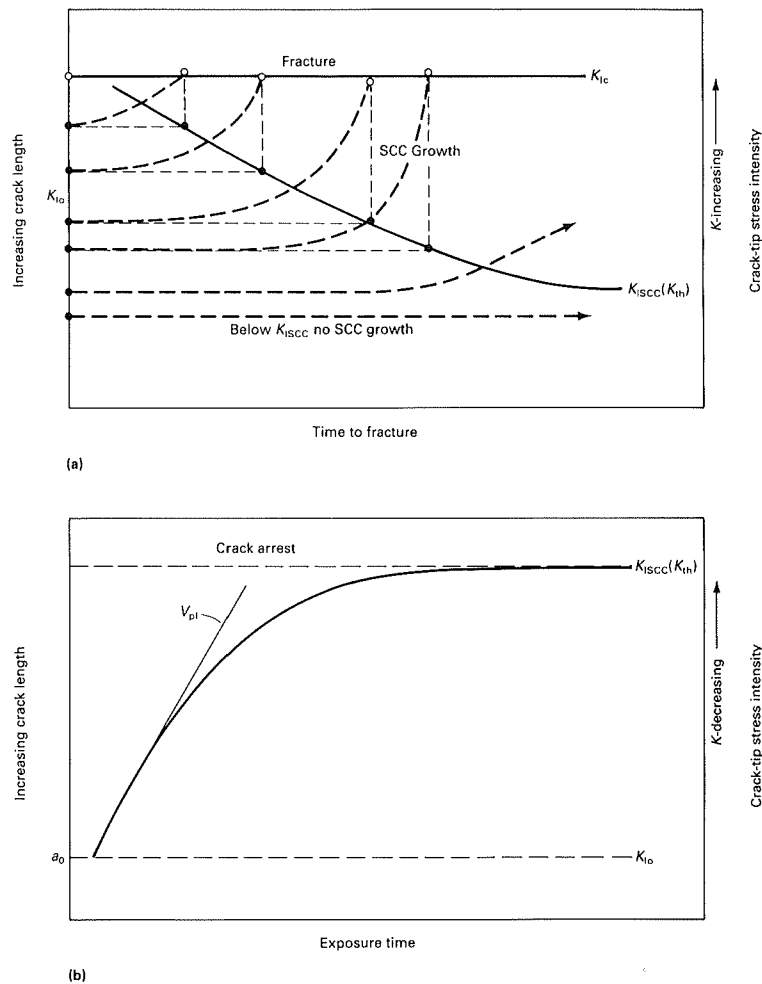


Fig. 19 Schematic comparison of determination of K_{ISCC} by crack initiation versus crack arrest. (a) Constant load test. (b) Constant initial crack opening displacement test. a_0 = depth of precrack associated with the initial stress-intensity K_{Io} ; V_{pl} = plateau velocity (Ref 10).

Figure 21 is a schematic of types of environmentally assisted fatigue crack growth behavior (corrosion fatigue) compared to growth in an inert environment (Ref 12). Environmentally-assisted fatigue crack growth may be broadly characterized in terms of the three general patterns of behavior illustrated. The first type, Type A, represents those material-environment systems in which the environmental effect results from the synergistic actions of fatigue and corrosion. Such behavior is observed in tests of aluminum in water. The environmental effect is evidenced by a reduction in the apparent threshold for crack growth and increases in the rate of crack growth at given ΔK levels. As ΔK , the fracture toughness value, approaches either K_C or K_{Ic} , the environmental influences are diminished as a result of either the rate-limiting nature of transport processes or other mechanical-chemical interactions.

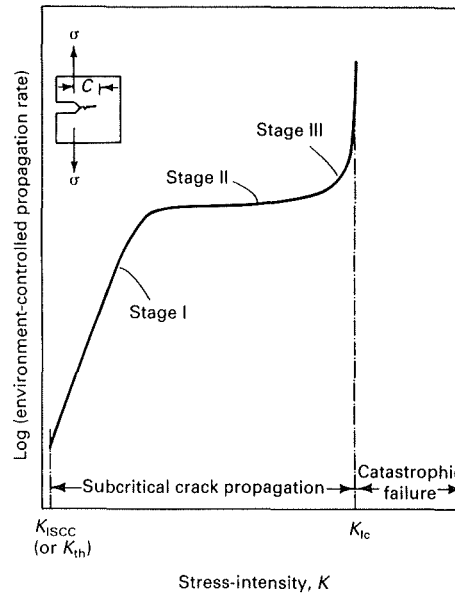


Fig. 20 Typical, subcritical crack propagation versus stress-intensity factor relationship. The stress-intensity factor, K , is defined as $A\sigma(\pi a/B)^{1/2}$, where σ is the applied stress, a is the crack length, and A and B are geometrical constants.

The second type of behavior, Type B, typified by the hydrogen-steel system, represents those systems where there is a substantial environmentally enhanced sustained load crack growth component. The environmental effects are quite strong above some frequency-dependent apparent threshold for stress-corrosion cracking, K_{ISCC} , and are negligible below this level. A broad range of material-environment systems exhibits behavior that falls between these two extremes, Type C. The response of materials will depend not only on the environment, but also on the loading frequency, the mean stress level, and waveform.

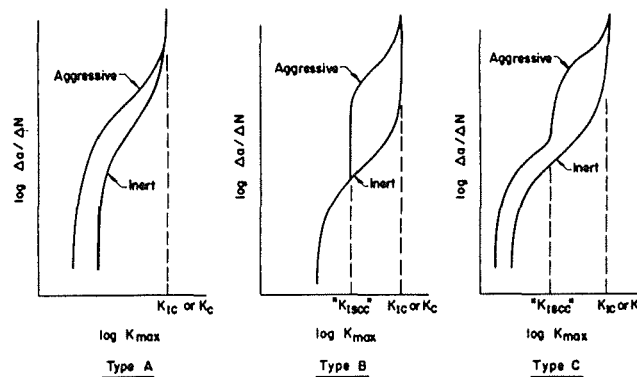


Fig. 21 Types of environmentally assisted fatigue crack growth behavior (Ref 12).

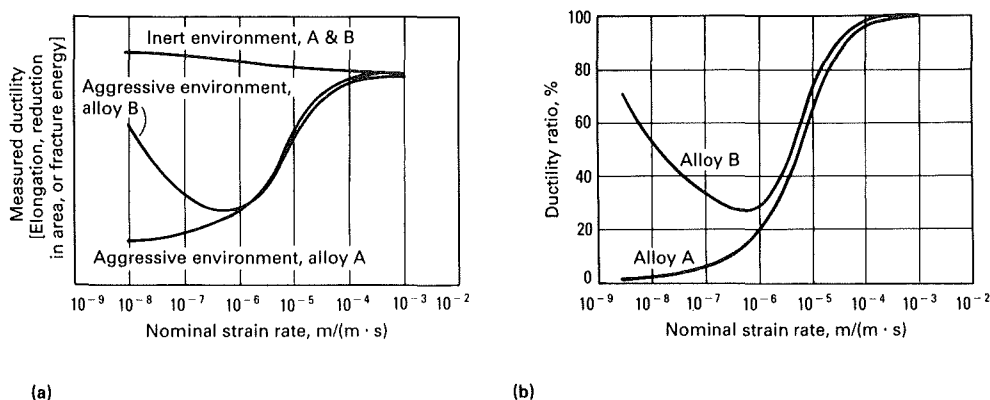


Fig. 22 Strain-to-failure plots resulting from slow strain rate testing. (a) Schematic of typical ductility versus strain rate behavior of two different types of alloys tested by the slow strain rate technique. (b) The ductility ratio is the ratio of a ductility measurement such as elongation, reduction in area, and fracture energy measured in the aggressive environment to that obtained in the inert reference environment.

Category 3: Smooth, Dynamically Loaded Specimens

This category includes primarily the slow strain rate specimens (SSRT) used in the evaluation of susceptibility to stress-corrosion cracking. The most recently developed method for accelerating the stress-corrosion cracking process in laboratory testing involves relatively slow strain rate tension testing of a specimen during exposure to appropriate environmental conditions. The application of slow dynamic strain exceeding the yield strength assists in the initiation of stress-corrosion cracking by film rupture in accordance with various general mechanisms for stress-corrosion cracking. A principal advantage of slow strain rate testing is the rapidity with which the stress-corrosion cracking susceptibility of a particular alloy and environment can be assessed.

Slow strain rate tests are carried out to fracture. Usually, a tensile testing machine pulls a sample that is exposed to the environment at a low crosshead speed (10^{-5} to 10^{-9} m/s). The strain-to-failure in the corrosive environment and the strain-to-failure in an inert environment can then be plotted against the strain rate, as shown in Fig. 22(a), or the ratio of these measurements can be plotted, as shown in Fig. 22(b). The schematic results for Alloy B indicate that at the lowest strain rates the rate of cracking is low because there is sufficient time for protective films to reform. As the strain rate is increased, there is a minimum in ductility that corresponds to a maximum in the rate of cracking. At still higher rates, there is insufficient time for corrosive attack, and the ductility rises. The ratio of other tensile property measurements, such as reduction in area and ultimate tensile strength, may be plotted. If the extension rate is controlled rather than the strain rate, the test may be called a constant extension rate test (CERT). They are sometimes also referred to as straining electrode tests. These are excellent tests for comparing the relative susceptibility of alloys to cracking in an environment or for studying the influence of metallurgical variables on the susceptibility of an alloy. However, the application of the data to the prediction of actual in-service crack propagation rates is difficult and unreliable (Ref 13).

Mechanisms for Stress-Corrosion Cracking

It has been pointed out (Ref 14) that a general mechanism for stress-corrosion cracking is an unreasonable and unattainable goal. Specific processes appear to operate under specific sets of environmental and metallurgical conditions. From a rather broad point of view, it can be said that there is a general mechanism in the sense that slip processes do operate, dissolution does occur, electrochemical reduction does occur. A montage showing many of these common processes is shown in Fig. 23. However, it is the degree to which each of these component processes operates that controls the cracking. Thus, in a given cracking process, only three or four component processes operating together may be necessary to specify the sufficient conditions for cracking. It is this critical combination of processes that comprises the "mechanism."

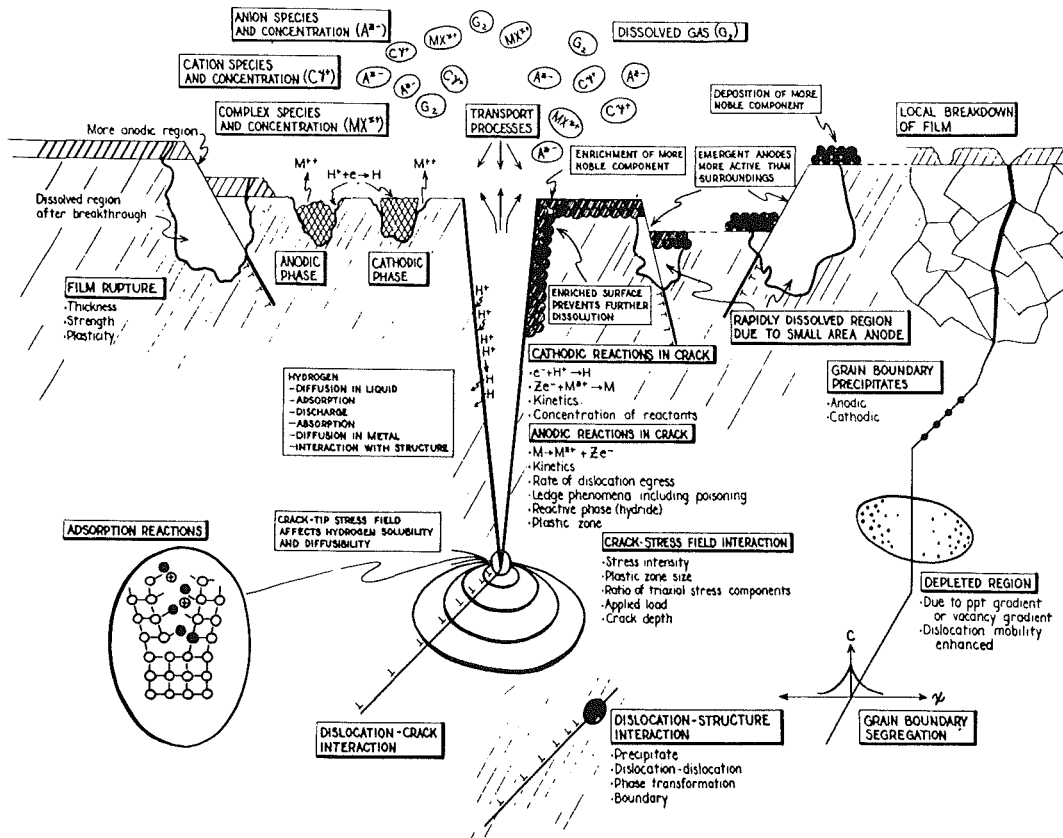


Fig. 23 Montage showing important processes that affect stress-corrosion cracking (Ref 14).

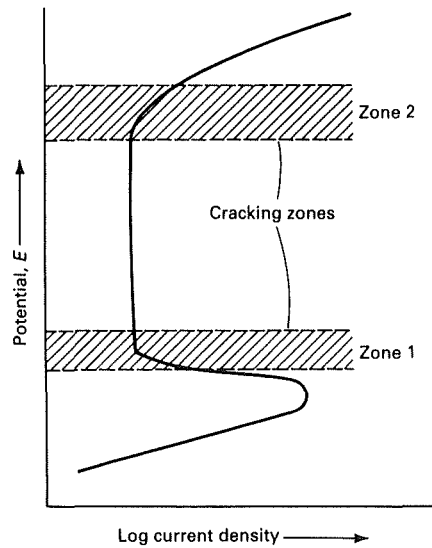


Fig. 24 Potentiokinetic polarization curve for an alloy-environmental system exhibiting passivation, and electrode potential values at which stress-corrosion cracking occurs.

The first step in stress-corrosion cracking is the localized breakdown of the surface oxide film. In many cases, this occurs by slip in the alloy and results in the exposure of unfilmed metal at the slip step. The localized dissolution of this exposed material, as influenced by the potential level, is significant in some mechanisms. For example, in the case of stainless steels, there are two potential regions wherein transgranular cracking is observed. These regions are indicated in Fig. 24. One of these, the lower zone in Fig. 24, is in the active-passive region; the other is in the transpassive region. The localized corrosion at the crack tip, which is needed for crack propagation, can occur in these regions, because films will form over the crack surfaces, thereby preventing general corrosion. In these potential regions, therefore, general corrosion and corrosive blunting of the crack tip does not occur, which is a general requirement of any stress-corrosion cracking mechanism. The role of films is significant in some of the proposed mechanisms, which are discussed below.

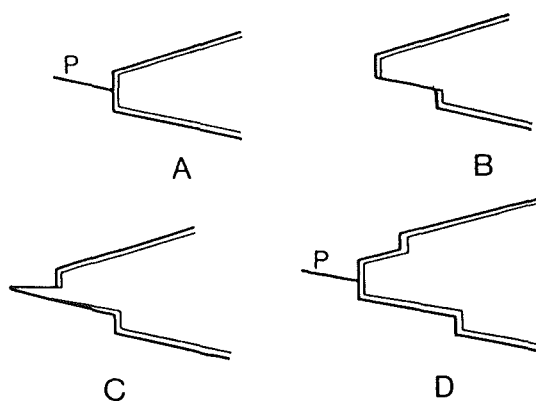


Fig. 25 Schematic diagram of the sequence of events occurring at the tip of a propagating stress-corrosion crack (Ref 15).

Repeated Film Rupture

A schematic diagram of the sequence of events occurring at the tip of a propagating crack is shown in Fig. 25 (Ref 15), and the sequence of events in going from A to D in this figure is as follows. (A) The tip of the crack and the surfaces of the crack are covered by a film. P is a slip plane at the crack tip. (B) The slip plane undergoes a shear and a new, unfilmed reactive surface is created as the crack tip moves forward. (C) Corrosive attack occurs on the reactive surface while repassivation begins on the outer edge. The morphological aspects of the attack may vary from alloy to alloy, and it is not necessary that the attack occurs preferentially along the slip plane. (D) Repassivation occurs and the process is ready to repeat, provided the strain rate at the crack tip is of sufficient magnitude to ensure repeated rupture of the film together with crack advance. A variation of this type of model applied to intergranular cracking of brass in a tarnishing solution is shown in Fig. 26 (Ref 16).

Slip-Dissolution

In the simplest view of this process, all that is required is that slip ruptures the surface oxide film and that dissolution along an active path then takes place to advance the crack. In precipitation-hardened aluminum alloys, there are electrochemically favorable paths for cracking along grain boundaries because of the presence of denuded zones and grain boundary precipitates. A process referred to as "active path dissolution" has been proposed to account for stress-corrosion cracking in these alloys. In this view, compounds such as $MgZn_2$, which are anodic with respect to their surroundings along a grain boundary, dissolve preferentially, and the applied stress acts to tear ligaments of remaining grain boundary material. Behind the crack tip, the crack surfaces are coated with a film so that general corrosion does not occur. In non-age-hardenable materials, these active paths are not present, and in the absence of such favorable paths for cracking, a process of stress-assisted dissolution may occur. The process can result in complete dissolution of the entire plastic zone at a stress raiser rather than cracking.

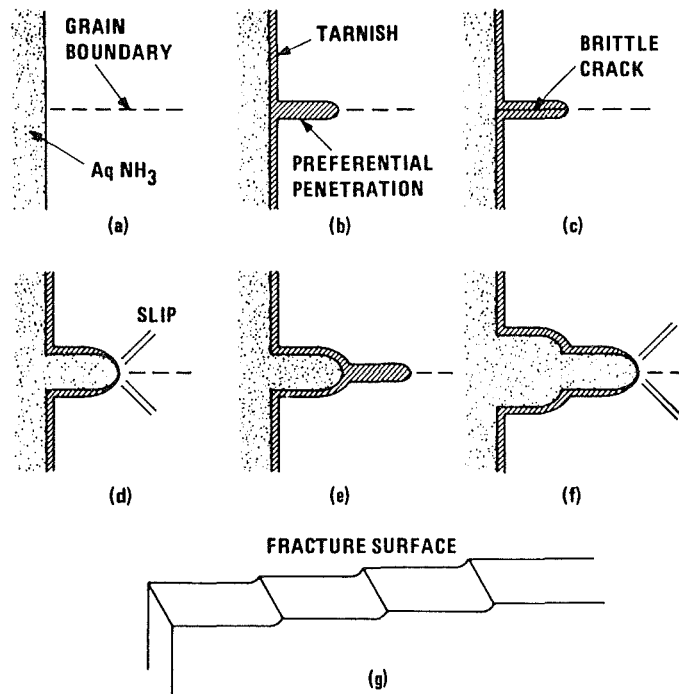


Fig. 26 Schematic illustration of the tarnish-rupture mechanism for stress-corrosion cracking of α -brass in tarnishing solutions. The tarnish is considered to grow preferentially along a grain boundary (b), until it undergoes brittle fracture (c). The crack is arrested by slip in the substrate, exposing metal to the environment (d). Further intergranular penetration then occurs (e), leading to further limited fracture (f). Crack propagation thus proceeds by repeated cycles, the resulting fracture surface showing evidence of discontinuous fracture (g) (Ref 16).

A slip-dissolution/film rupture mechanism was chosen to model stress-corrosion cracking in types 304 and 316L stainless steel and A533B and A508 low-alloy steel in water at 288 °C (Ref 17). In this model (Fig. 27), crack advance is related to the oxidation reactions that occur at the crack tip as the protective film is ruptured by increasing strain in the underlying matrix. This rupture event occurs with a periodicity, t_f , which is determined by the fracture strain of the oxide and the strain rate at the crack tip. In this model, fracture of the film is not contingent upon the occurrence of plastic deformation in the substrate, as in other models. The extent of crack advance is related by Faraday's Law to the oxidation charge density associated with dissolution and oxide growth (passivation) of the bared metal surface, as shown in Fig. 27. The crack growth rate, V_t , can be given in general form as:

$$V_t = f(n)(\dot{\epsilon}_{ct})^n \quad (5)$$

where $\dot{\epsilon}_{ct}$, the crack tip strain rate, embodies the mechanical contributions, and n is a parameter that represents the effects of the environment and material chemistries on environmentally assisted crack growth.

There are limits to the validity of the crack growth rates predicted by the above equation, as indicated in Fig. 28. At high crack tip strain rates of the order of $10^{-2}/s$, a bare surface is maintained continuously at the crack tip, and the environmentally assisted crack growth rate becomes independent of the crack tip strain rate because it cannot exceed the Faradaic equivalent of the bare surface dissolution rate. At the very lowest crack tip strain rates, sharp cracks are not maintained when the average crack tip propagation rate approaches the oxidation rate on the crack sides. Under this particular condition, the crack propagation rate decreases with exposure time and crack arrest occurs due to blunting.

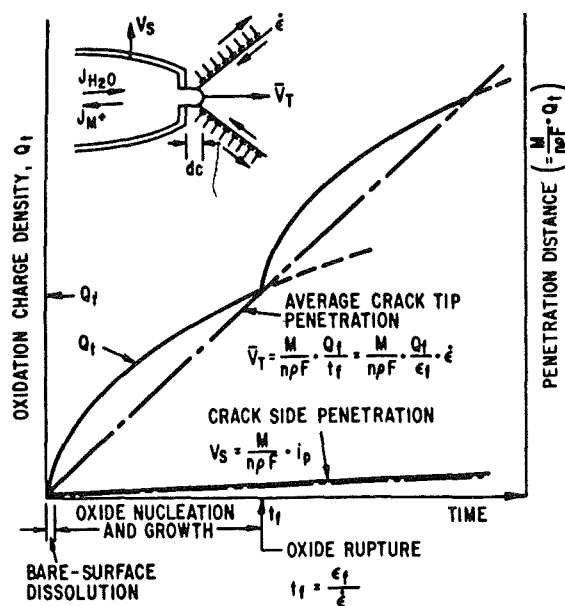


Fig. 27 Schematic oxidation charge density/time relationships for a strained crack tip and unstrained crack sides (Ref 17).

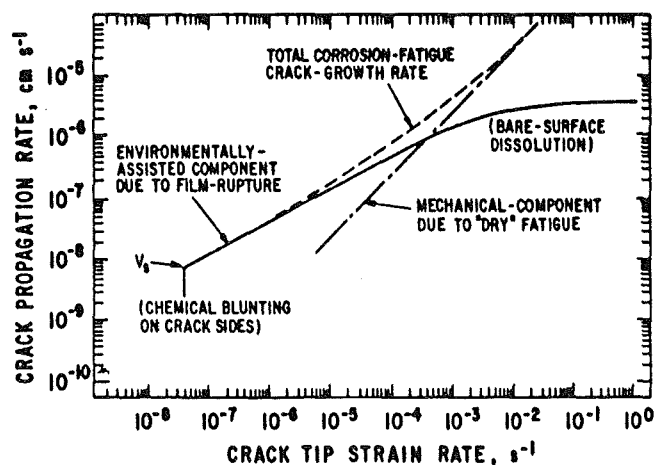
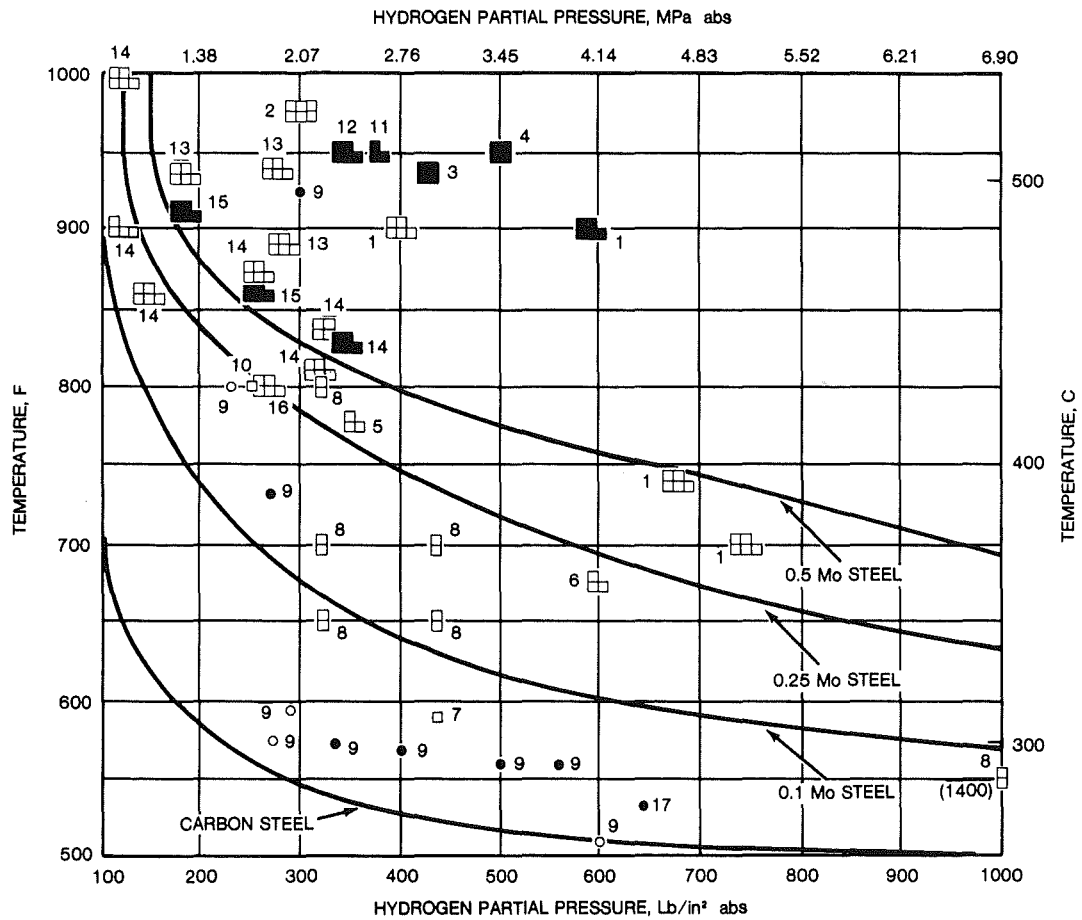


Fig. 28 Strain rate dependence of the crack growth rate for mechanical (fatigue) and environmental (due to film rupture/slip dissolution) components of crack advance (Ref 17).

Hydrogen-Assisted Cracking

Much current research in both stress-corrosion cracking and corrosion fatigue emphasizes the importance of hydrogen in these processes, although the specifics of how hydrogen assists the cracking process are still open to debate. There is evidence that, in aluminum alloys, steel, titanium alloys, and in nickel-base alloys, hydrogen-assisted cracking is an important aspect of the overall cracking process. In this Atlas, we are concerned about the influences of hydrogen in stress-corrosion cracking and corrosion fatigue.



References	Analysis Mo Equivalent			LEGEND
	CR	Mo	V	
1. Shell Oil Co.*		0.50	0.50	EQUIVALENT MOLYBDENUM CONTENT 0 0.01-0.10 0.11-0.20 0.21-0.30 0.31-0.40 0.41-0.60 0.61-0.80
2. Weld Deposits, D.J. Bergman*	0.79	0.39	0.59	
3. Weld Deposits, D.J. Bergman*	0.80	0.15	0.35	SATISFACTORY HYDROGEN ATTACK
4. Weld Deposits, D.J. Bergman*	0.50	0.25	0.37	
5. Continental Oil Co.*		0.25	0.25	0 (open circle) 0.01-0.10 (open square) 0.11-0.20 (open rectangle) 0.21-0.30 (open cross) 0.31-0.40 (open diamond) 0.41-0.60 (open star) 0.61-0.80 (open hexagon)
6. Standard Oil Co. of California*		0.27	0.27	
7. Standard Oil Co. of California*	0.05	0.06	0.08	● (solid circle) ■ (solid square) ▩ (solid rectangle) ⊠ (solid cross) ◆ (solid diamond) ☆ (solid star) ☆ (solid hexagon)
8. A.O. Smith Corp.*		0.13-0.18	0.11	
9. Shell Development Co. Drawing No. VT 659-2,				NOTES:
10. AMOCO Oil Co.*	0.04		0.01	
11. R.W. Manuel, <i>Corrosion</i> , 17 [9], pp. 103-4, Sept., 1961	0.27	0.15	0.22	1. Mo has four times the resistance of Cr to H ₂ attack.
12. The Standard Oil Co. of Ohio*	0.11	0.43	0.50	2. Mo is equivalent to V, Ti, or Nb up to 0.1 percent.
13. Exxon Corp.*				3. Si, Ni, Cu, P, and S do not increase resistance.
14. Union Oil of California*				
15. Amoco Oil Co.*				
16. Standard Oil Co. of California*				
17. Gulf Oil Corp.*				

*Private communication to Subcommittee on Corrosion

Fig. 29 Operating limits for steels in hydrogen service showing effect of trace elements (Ref 18).

However, at elevated temperatures there are problems of a different nature, because steels in a hydrogen environment can undergo decarburization, as well as blistering due to the formation of CH_4 at internal pores. To avoid these problems, operating limits have been set for steels in hydrogen service, as indicated by the Nelson diagram in Fig. 29 (Ref 18).

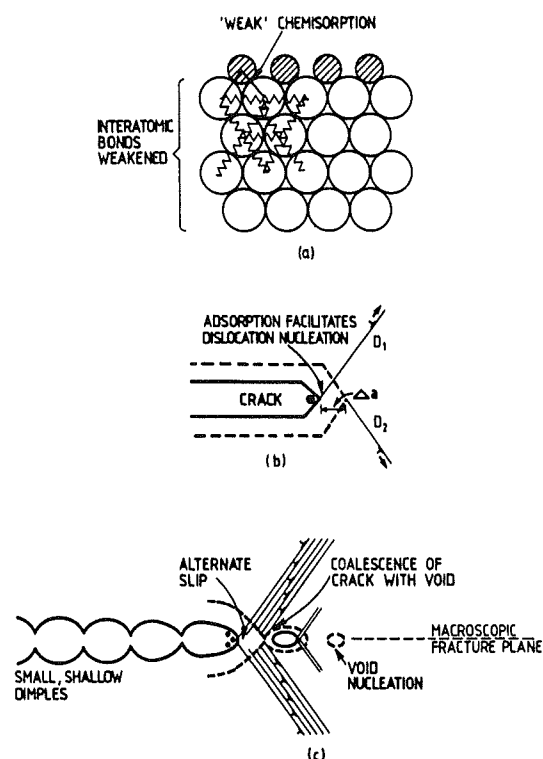


Fig. 30 Schematic diagrams illustrating the adsorption-induced localized slip process for environmentally assisted cracking (Ref 21).

At room temperature, the problems associated with hydrogen are related to embrittlement. For example, it is known that charging a notched steel specimen with hydrogen can lead to delayed failure of a brittle nature, as shown in Fig. 18. It has been proposed that the volume expansion associated with the state of triaxial stress ahead of the notch or crack results in the diffusion of carbon to this region in sufficient concentration to weaken enough iron-iron bonds to develop localized rupture. However, it now appears that the increase in hydrogen concentration due to triaxiality at a notch is not sufficient for the concentration of hydrogen to increase enough to bring about embrittlement (Ref 19). In view of this finding, it has been postulated that localized regions of extremely high triaxiality do exist and allow the hydrogen level to increase to a level sufficient to result in localized rupture (Ref 20).

Another view that is supported by much recent evidence is that hydrogen adsorption at a crack tip or notch promotes the localization of the plastic flow and microvoid coalescence processes (Ref 21). A schematic of this model is shown in Fig. 30. In Fig. 30(a), adsorbed hydrogen atoms weaken interatomic bonds and thereby facilitate the shear movement of atoms (i.e., dislocation nucleation) at crack tips. In Fig. 30(b), dislocation injection from the crack tip first on plane D_1 and then on plane D_2 produces an increment of crack advance, Δa . In Fig. 30(c), some dislocation activity (not shown) occurs ahead of the crack tip producing voids. Cracks, growing by alternate slip, coalesce with those voids that are formed along the line of prolongation of cracks, so that the macroscopic fracture plane bisects the active slip planes.

Another possible role of hydrogen is to form hydrides, as has been observed in titanium, for example. These hydrides have weak interfaces with the surroundings, and it is reasonable to expect that localized rupture can occur along such interfaces.

Models for Corrosion Fatigue

With respect to the modeling of corrosion fatigue, most approaches have dealt with crack propagation rather than the crack initiation stage. Processes such as film rupture, dissolution, and hydrogen-assisted cracking, which are important in stress-corrosion cracking, are also important in corrosion fatigue. One approach to the analysis of corrosion fatigue crack propagation behavior is to assume that the effect of the environment can be superimposed upon the purely mechanical aspects of the process. In this approach, the rate of fatigue crack propagation is the sum of two components:

$$da/dN = A\Delta K^m_{\text{mech}} + B_{\text{env}} \quad (6)$$

This equation was used in the analysis of the rate of fatigue crack growth of brass in an ammoniacal copper sulfate environment (Ref 22). B is the amount of tarnish formed per cycle, a function of frequency. The thickness of the tarnish film, x , as a function of time, t , in the absence of cyclic loading, was experimentally found to be:

$$x = 27.9 \times 10^{-4} t^{1/2} \quad (7)$$

where x is in millimeters and t is in hours. The results of the analysis, are shown in Fig. 31. In this analysis, it was assumed that the fatigue behavior in air was independent of frequency.

A superposition approach has also been employed in the analysis of the effects of water vapor or aqueous environments on the corrosion fatigue crack propagation in high-strength alloys where hydrogen is thought to exert an important environmental influence (Ref 23). The processes involved are:

1. Transport of the deleterious species to the crack tip
2. Reactions with the newly produced crack surface to produce hydrogen or to effect dissolution
3. Hydrogen entry (or absorption)
4. Diffusion of hydrogen to the fracture (or embrittlement) site
5. Partitioning of hydrogen among the various microstructural sites
6. Hydrogen-metal reactions leading to embrittlement (that is, the embrittlement reaction)

The rate of fatigue crack growth was expressed as:

$$da/dN = (C_0 + C_e/f)\Delta K^2 \quad (8)$$

where C_0 is a coefficient for crack growth in an inert environment, C_e is a coefficient for the added crack growth in the aggressive environment, and f is the cyclic frequency. The expression was found to be in accord with experimental results.

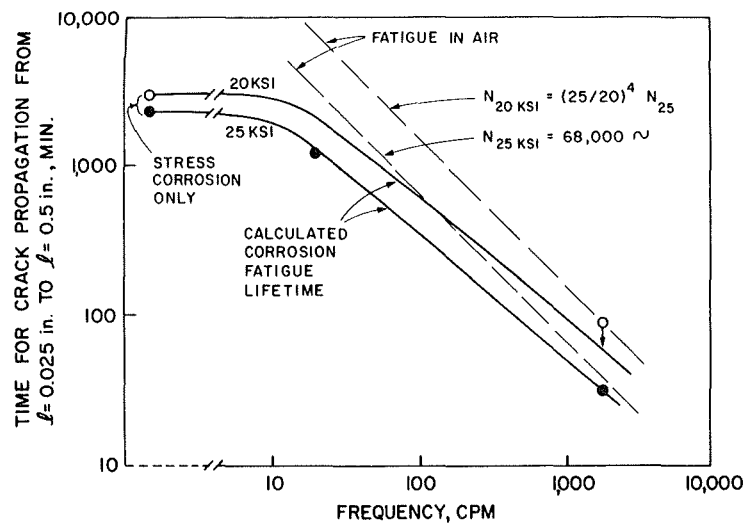


Fig. 31 Time required to propagate a fatigue crack in cold rolled 70-30 brass from a semi-length, l , of 0.025 in. to a semi-length, l , of 0.5 in., in air and in an ammoniacal copper sulfate environment as a function of frequency (Ref 22).

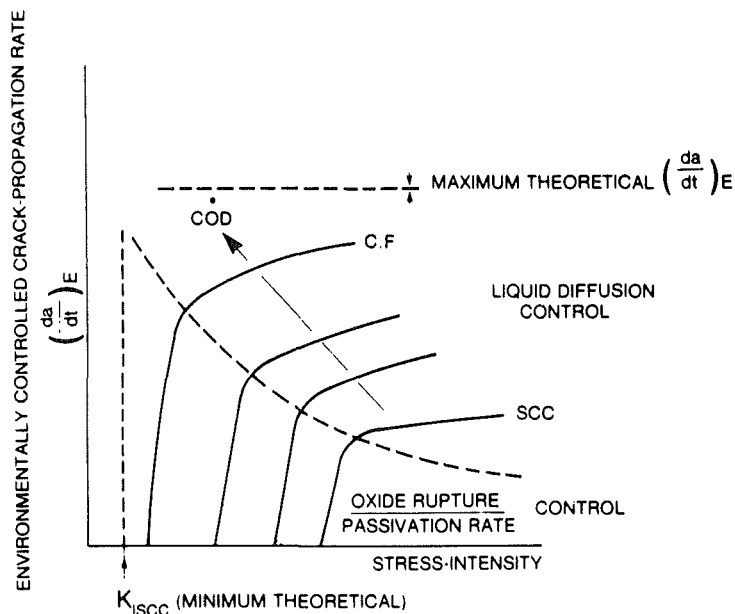


Fig. 32 Schematic variation of the crack propagation rate with stress-intensity for different instantaneous crack tip strain rates. Note the smooth transition between stress-corrosion cracking and corrosion fatigue regimes associated with different rate-determining processes (Ref 24).

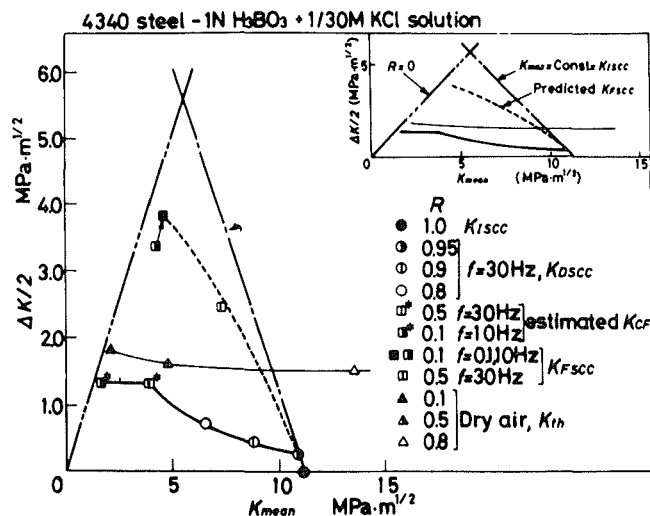


Fig. 33 Endurance limit diagram for 4340 steel in 1N $\text{H}_3\text{BO}_3 + 1/30\text{M}$ KCl solution at 298 K (Ref 25).

A generalized treatment of corrosion fatigue and stress-corrosion cracking has developed, which is based upon Eq 5 (Ref 24). From considerations of the effect of strain rate in crack propagation in ductile alloy/aqueous environment systems, a smooth transition in da/dt values with changes in the instantaneous strain rate is expected, and such a transition has been noted for sensitized stainless steel in water under

static, rising, and cyclic loading conditions. It follows therefore that any commonly quoted values of threshold stress (σ_{th}) and stress-intensity (K_{Isc}) for the onset of the environmental enhancement of cracking apply only to static conditions, and if there is any superimposed dynamic load, the σ_{th} or K_{Isc} will decrease, as has been observed in stainless steel/water and low-carbon steel/bicarbonate systems. Figure 32 indicates the expected variation of the crack propagation rate with stress-intensity for different instantaneous crack tip strain rates.

In the development of a general model for corrosion fatigue crack propagation, the role of mean stress also needs to be considered. It has been proposed that there is a boundary, indicated by the dashed line in Fig. 33, between a dynamic stress-corrosion cracking region and a corrosion fatigue region (Ref 25). Above this dashed line, intergranular fractures that are characteristic of stress-corrosion cracking in 4340 steel were found, and below the dashed line, transgranular fractures associated with corrosion fatigue, were found. The threshold values in air and in the test solution are also shown in Fig. 33.

In the foregoing discussion of corrosion fatigue, we have been concerned with the effects of water vapor and aqueous environments. Oxidation can also exert an important environmental influence on corrosion fatigue behavior, particularly at elevated temperatures in air, and sometimes with unexpected results. For example, under continuous cycling, 2.25Cr-1Mo steel tested in air exhibited a decrease in life compared to those obtained in vacuum, as expected. However, if a 10-min hold time is introduced into each cycle in air, either at the maximum tensile strain or at the maximum compressive strain, it was somewhat unexpected to find that the hold in compression shortened the life more than a hold in tension. At a total strain range of 0.85 at 593 °C, the fatigue lifetime with a tension hold was more than 1065 cycles, whereas with a compression hold the lifetime was only 799 cycles. The reason for this behavior was due to the oxide. The oxide that formed during a tension hold spalled off the surfaces of the specimen during subsequent compression. However, the oxide that formed during compression did not spall off in the subsequent tensile portion of the cycle. Instead the oxide cracked, and the formation of these cracks resulted in a shortening of the fatigue life relative to the tension hold case (Ref 26).

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