

NRL Report 6041

UNCLASSIFIED

STRESS-CORROSION CRACKING AND RELATED PHENOMENA IN HIGH-STRENGTH STEELS

A REVIEW OF THE PROBLEM, WITH AN ANNOTATED BIBLIOGRAPHY

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November 6, 1963



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ABSTRACT

The problem of the stress-corrosion cracking of high-strength steels is contrasted with stress corrosion in other metals. The implications of hydrogen embrittlement and especially of brittle fracture to interpretations of stress-corrosion cracking tests on high-strength steels are summarized. Test methods are briefly described, and the results of recent studies as reported in the literature are summarized topically. A discussion of the meaning of these findings is given, together with recommendations regarding the further development of test concepts. An annotated bibliography of thirty-six entries is appended.

PROBLEM STATUS

This concludes the phase of the problem dealing with the review of the stress-corrosion cracking of high-strength steels. Work on other phases is continuing.

AUTHORIZATION

NRL Problem M04-02 (R05-24G)
Bureau Project WW 041, Task R05-24G
Some of the studies made by the author
on this phase were supported by Advanced
Fleet Deterrent funding.

Manuscript submitted November 6, 1963.

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INTRODUCTION

The term stress-corrosion cracking (SCC) is used here for the process in which a crack propagates by the stress-induced corrosion of metal at the advancing tip of the crack. There is much evidence that this corrosion is not a homogeneous reaction (in which the exchange of electrons between reacting species occurs locally, on an atomic scale) but rather an electrochemical one, in which a small electric current flows over distances which are large, on an atomic scale. Part of the electrical circuit is in the metal, and the current takes the form of a flow of electrons, and part of it is in the corrodent, and the current takes the form of a drift of ions. This is depicted in simplified form in Fig. 1. The reader interested in more detailed reviews of the theory of SCC, insofar as it has been worked out on metals other than the high-strength steels, is referred to Refs. 1 and 2.

The fact that many important metals, including stainless steels and high-strength aluminum alloys, are subject to SCC under certain conditions has long been known. Either these metals are not used where such cracking might occur, or the conditions of stress, heat treatment, and environment are carefully attended to in order to minimize the likelihood of cracking. Even when SCC occurs in alloys such as type 304 stainless steel or type 2024 aluminum alloy, the results can hardly be described as spectacular. The finale to the process of SCC in an austenitic stainless steel pressure vessel, for example, is usually simply a leak. This leak may be annoying, costly, or hazardous, depending on the nature of the effluent; but even if the leak is a calamity, it is in general a quiet calamity.

Largely during the 1950's it became evident that some, and perhaps many, of the high-strength steels being used in large structures, or contemplated for such structures, were also susceptible to SCC.* The finale in the SCC process in these steels, unlike the case of austenitic stainless steels or the nonferrous metals, is catastrophic, and by way of example the result in a pressure vessel made of one of them is not a leaking vessel but a shattered one. These failures are apt to be all the more newsworthy because of the nature of the structures which are being made of these steels, or which designers would often like to make out of them: large rocket motor cases, control rods for nuclear reactors, hulls for advanced deep-diving submersibles, skins and landing gear for high-speed aircraft, and autoclaves for high-pressure chemical processes. There have been therefore in recent years projects seeking to identify those high-strength steels which are susceptible to SCC in various environments and to try to determine the conditions of heat treatment, stress, and environment under which an otherwise susceptible steel could be safely used. The present report is intended to summarize the problem and the findings of these projects.

This review was separated from an earlier review of the problem in austenitic stainless steels (3) because cracking in the two classes of steels differs in at least four important particulars.

*The term "high-strength steel" is used in this report to include martensitic, precipitation-hardening, and maraging steels, but not cold-worked austenitic steels.

1. The austenitic steels rarely crack at room temperature, whereas the high-strength steels do so readily.
2. Specific ionic species (halides or hydroxides) are believed necessary to crack the austenitic steels, but these species are not required to crack the high-strength steels.
3. The austenitic steels are not susceptible to hydrogen embrittlement cracking (HEC) in the same sense that the high-strength steels are.
4. The high-strength nonaustenitic steels are notch sensitive, whereas the austenitic steels are notch tough. This is a simplified qualitative statement of a complex phenomenon of major importance to the technology of high-strength steels.

Points 1 and 2 are self-explanatory. The present review includes a discussion of points 3 and 4 in a section on mechanisms of cracking. This is followed by a summary of methods commonly used in testing for SCC in high-strength steels. Some of the more important laboratory findings to date are briefly summarized, and this summary of findings is followed by a concluding discussion.

There is appended an annotated bibliography of recent papers and reports (from 1950 onward) on the subject which the reader interested in original material may find useful. No pretense is made that this is anywhere near a complete bibliography even of only the U.S. literature bearing on the subject, for such completeness would require an enormously larger effort than is felt justified. (It would have required sifting all the literature on sulfide cracking and on hydrogen embrittlement, for example.) Nevertheless the bibliography is believed to contain most of the papers and reports which present original ideas on the fundamentals of the problem and also most of those reporting extensive systematic testing. One of the entries in the bibliography has nothing at all to do directly with SCC but is included because it contains data on the susceptibility of some important high-strength steels to HEC.

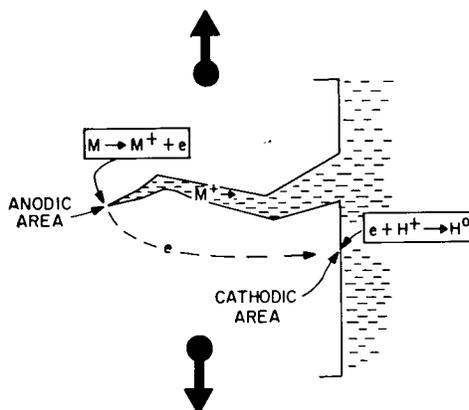
MECHANISMS OF CRACKING

Stress-Corrosion Cracking

There is much disagreement about the mechanisms by which SCC occurs in metals in general, and indeed there is incomplete agreement on the question of which cracking phenomena should be termed stress-corrosion cracking. There seems no present justification to complicate an already complex problem by attempting to arbitrarily fit it into a common frame embracing such other phenomena as brittle fracture and cracking by liquid metals, and accordingly SCC will be reserved here for those cracking phenomena necessarily requiring stress-induced electrochemical corrosion of metal at the advancing tip of the stress-corrosion crack. The stress may be either externally applied or "locked in" (from welding or heat treatment, for example). The corrodent need not be a bulk aqueous solution - it may be a thin adsorbed layer of an aqueous solution, or it may be nonaqueous, as a molten salt.

When these conditions are properly satisfied in a susceptible material, metal goes into solution at the advancing tip of the crack, which by definition is therefore an anode. Simultaneously electrons are released into the metal, through which they flow to another discrete area of the metal also in contact with the electrolyte. At this second area the electrons participate in a reduction reaction, which by definition makes that area a cathode. The cathodic (reduction) reaction may be the reduction of H^+ to H^0 , which may react with dissolved oxygen to form water, or may form bubbles of hydrogen gas, or may dissolve in the metal. This last possibility is important to the fundamentals of the cracking of high-strength steels, as will be noted later.

Fig. 1 - Stress-corrosion cracking (schematic)



In Fig. 1 the cathodic area is shown outside the crack. In some cases there is no reason the cathodic reaction cannot occur deep within the crack, and in fact this has been demonstrated to occur (4).

The propagation of the stress-corrosion crack depends upon the flow of current between the anodic and cathodic areas. If the electrochemical potential of the cathodic areas were adjusted to the potential of the anodic areas, then the driving force of the electrochemical process would be reduced to zero, no current would flow, and the cracking would be expected to stop. This manipulation of the potential can be done by impressing small currents from an external source, and the act is called cathodic protection. This can be effective in stopping or at least retarding cracking, as has been demonstrated in several systems (see, for example, Ref. 5). Cathodic protection and some concomitant side effects of practical importance to high-strength steels will be touched on again in later sections.

Almost all of the theory of SCC insofar as it has been worked out to date has derived from experiments with metals other than high-strength steels. It seems unwise to assume at this stage that all details of the theory must necessarily apply to the high-strength steels until experience justifies it. For this reason no further details of SCC theory will be recounted here.

Hydrogen Embrittlement Cracking

If a steel is sufficiently strong and not too tough, and if it contains more than some minimum amount of hydrogen, a sustained tensile load may cause a crack to nucleate and slowly grow at a stress far smaller than that required to cause fracture in the absence of hydrogen. This is one manifestation of hydrogen embrittlement, and it is also known under such names as delayed fracture and static fatigue, among others. Data showing the delayed fracture of type 4340 steel as a function of stress are shown in Fig. 2.

Lower ductility, higher hydrogen content, and greater degrees of triaxiality of stress speed this cracking process or diminish the load required to effect it. Although HEC was explicitly described in the technical literature 28 years ago (7), the basic mechanisms involved are still poorly comprehended. But there is no uncertainty about its serious practical implications to the use of the high-strength steels if hydrogen can get into them.

In common with the special case of SCC, all "wet" corrosion of steel is believed to occur by an electrochemical process in which the oxidation of iron at local anodic areas must be balanced at other areas by an equivalent amount of a reducing reaction, which may be the reduction of hydrogen. If this enters the steel, and if this is a high-strength

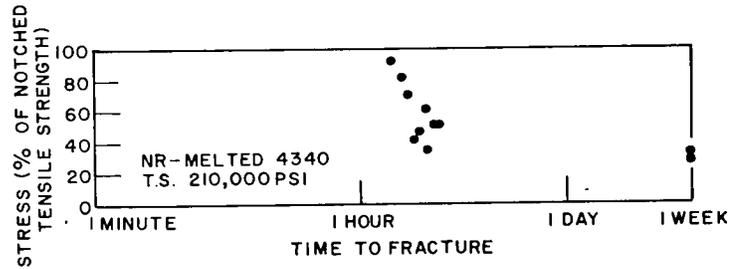


Fig. 2 - Delayed fracture of steel containing hydrogen from a cadmium electroplating process (6)

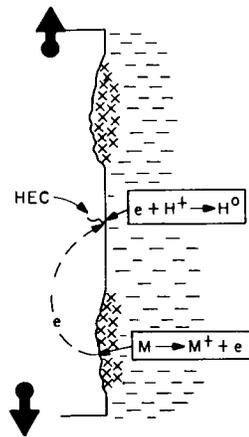
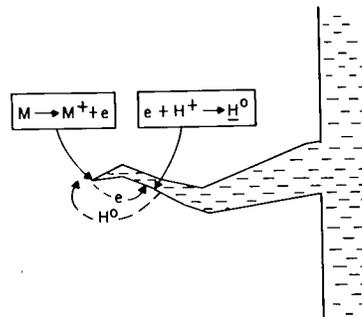


Fig. 3 - Hydrogen embrittlement cracking from aqueous corrosion (schematic, conjectural)

Fig. 4 - Mixed cracking mechanisms (SCC and HEC)(schematic, conjectural)



steel under sustained tensile loading, the steel may experience hydrogen embrittlement cracking. This possibility is illustrated schematically in Fig. 3. There appears to be no reason why this cathodic reaction cannot occur deep within a genuine stress-corrosion crack in some cases, so that both mechanisms of cracking may occur simultaneously or alternately (Fig. 4).

If one attempts to prevent corrosion of steel in an aqueous solution by applying cathodic protection, he finds that the potential of the steel must be depressed below the potential at which hydrogen can be reduced from the solution. This hydrogen may be reduced only slowly, or it may react with dissolved oxygen or pass off as bubbles; but

it also may enter the steel instead, to cause HEC. In fact, one of the standard laboratory methods for studying HEC in steels is to "cathodically protect" them (or, as commonly expressed, to "cathodically charge" them) in an aqueous solution. This cathodic charging of hydrogen may be done by an impressed current system, as from a battery, but it may also be done in many solutions including sea water by coupling the steel to a less noble metal, such as zinc or magnesium. (Fortunately the steel of which conventional ship hulls are made is not susceptible to HEC in the same sense that the high-strength steels are, and we can therefore cathodically protect them against corrosion with great economic advantage.)

Brittle Fracture

A metal is sometimes described as being notch sensitive if it fractures at a disproportionately low nominal stress in the presence of a notch, and the resulting break has come to be known as brittle fracture. One example of such a brittle fracture is shown in Fig. 5. This is a failure in a rocket motor case originating at a thumbnail-shaped area and running away from it in both directions. This fracture surface consists of at least three separate areas: the thumbnail itself, the flat brittle-fracture area (also sometimes called transverse fracture because it is transverse to the maximum principal tensile stress), and the "silky" borders called shear lips at an angle of approximately 45 degrees to the transverse fracture. The flat transverse fracture, if it has a long run, is characterized by chevron markings pointing toward the origin. Brittle fractures of this type propagate at high speeds - approximately 1/3 the speed of sound - and they are not affected by environment, as far as is known. In other words, such fractures are purely mechanical, and the appearance of either the chevrons (which may not always be clearly evident) or the shear lips (which may be so thin in very brittle steels that a hand lens is needed to see them) is sufficient to identify the fracture, or at least the portion of it so marked, as being other than stress corrosion or hydrogen embrittlement. Note in Fig. 5 the relatively small stress-corrosion area which set off the catastrophic failure. This is characteristic of the high-strength steels.



Fig. 5 - Thumbnail-shaped SCC flaw which triggered brittle fracture in a rocket motor case being hydrostatically tested. Note chevrons pointing to the origin of fracture. White borders are shear lips.

The theory and practice of evaluating the susceptibility of high-strength steels to brittle fracture is currently in a state of flux, but one method which is at least instructive is to insert a sharp crack in a strip specimen and then pull it to failure in a tensile machine. The higher the nominal stress at fracture, the tougher is the steel. One of the characteristics of the high-strength steels is that if such specimens are broken over a range of temperatures, there is a transition from brittleness at low temperature to toughness at high temperature. Data illustrating this is shown in Fig. 6. Two curves are shown in this figure representing two different conditions of heat treatment. Note that at about room temperature this steel is very tough in one heat treatment and very brittle in the other.

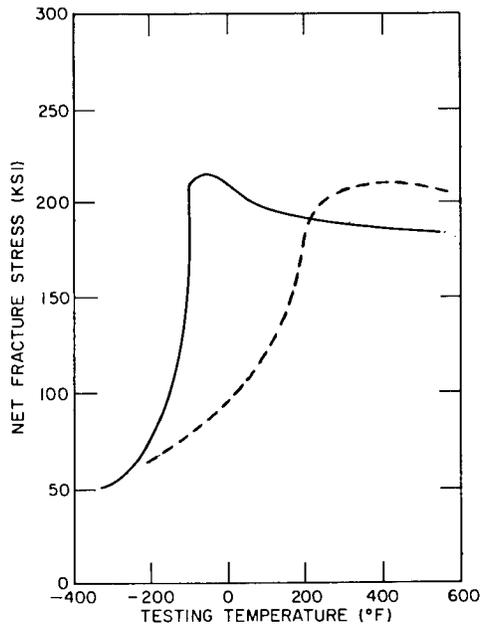


Fig. 6 - Effect of temperature on the net fracture stress of modified type 422 steel tempered at 900°F (broken curve) and 1050°F (continuous curve) (8). The centered-cracked strip specimens were 0.040 inch thick.

The tougher the steel, the larger is the "flaw" required to cause a brittle fracture to run under a fixed load. Also, a thick specimen is more susceptible to brittle fracture than a thin one, for purely geometric reasons; or, in other terms, the transition temperature for a thick specimen is higher than that of a thin specimen.

The foregoing is a highly abbreviated summary of brittle fracture technology. Three implications of this of particular importance to SCC of high-strength steels are as follows:

1. If one steel is more notch-sensitive (under the test conditions) than another, a small stress-corrosion crack (or even a pit or crevice corrosion area) may trigger brittle fracture in the notch-sensitive steel, whereas a much longer stress-corrosion crack might be necessary to cause the tougher steel to rupture. Thus the more notch-tough steel may require a longer time to fracture not because of a slower rate of SCC but because the SCC process must produce a longer crack before it is of the critical size to trigger the terminal, purely mechanical, fracture.

2. From purely geometric causes, a stress-corrosion crack may not have to be nearly as extensive to set off brittle fracture in a thick specimen as in a thin one.

3. Because of the transition temperature effect, the length of crack required to trigger brittle fracture is increased as the test temperature is increased in the transition range, which for many of the steels of interest and for many common specimen geometries may be in the vicinity of room temperature. If one attempts to measure a temperature coefficient of SCC by noting the effect of temperature on the time to failure, he may in some cases be measuring the combined effects of rising temperature on true SCC rates (which increase with increasing temperatures at least in the SCC of austenitic stainless steels) and on purely mechanical fracture, which tends to require a longer stress-corrosion crack and therefore a longer total time to fracture at higher temperature.

From the above, it is clear that both HEC and brittle fracture can play an important role in SCC tests of high-strength steels, whereas they can be safely ignored in SCC tests of most other metals, including the austenitic stainless steels.

EXPERIMENTAL METHODS USED IN SCC TESTS

Three methods of stressing specimens have been used:

1. Dead-loading in tension by a lever and weight system. This has the advantage of giving a much simpler stress pattern than the bent beam methods described below; and it offers, at least in principle, the possibility of observing the nucleation and growth of a number of cracks over a large area experiencing the same stress. Its principle disadvantage is the cost both in equipment and in space required, so that it will probably continue to be used, as in the past, largely for laboratory study of fundamental phenomena.

2. Spring-loading in tension using either helical springs or proving-ring-type springs. This method has the advantages of the dead-loading method with less demands, but is little used. Both the dead-loading and spring-loading methods will be recognized to be externally loaded, whereas the following method employs the specimen to supply its own stress.

3. Stressing bent-beam specimens, by any of six arrangements, as shown in Fig. 7. Customarily the bent beam specimens are "self-stressed," that is, the specimens are bent in a fixture so that the operating stresses in the SCC test are provided by the elastic stresses in the specimen itself. These stresses of course tend to relax during the cracking process. With the exception of the U-bend specimen, these specimens are usually stressed below the yield, and stresses are estimated from elastic theory. (Careful analysis has shown that the stress in a real bent beam specimen is highly complex, in particular that the maximum stress in the specimen of Fig. 7a is not at the top of the bow but at two points well down toward the ends.) The four-point loaded specimen has the advantage of stressing all the area between the two inner knife edges to the same stress, at least according to simplified analysis. (Another scheme for providing constant stress over extended area of specimen under bending load is one in which a cantilever beam specimen is tapered according to such dimensions that the stress remains constant over the test section.) The incomplete circle specimen of Fig. 7d, sometimes designated as a C-clip, is of interest to airframe manufacturers because of the simplicity of making the specimen from structural tubing, which is made of high-strength steels and used as an important structural form in airframes.

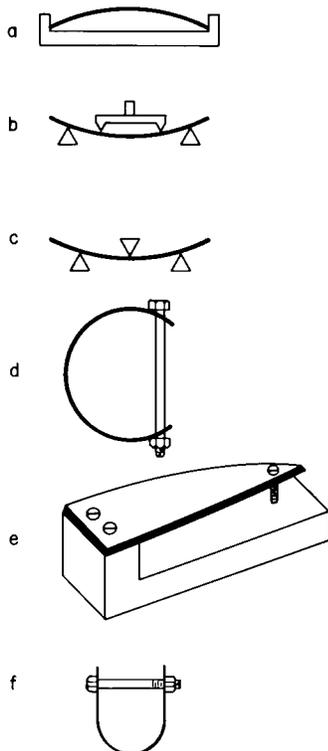


Fig. 7 - Methods of stressing strip specimens

There is no one corrodent which occupies the position in SCC tests of high-strength steels that boiling 42% $MgCl_2$ occupies in similar tests of austenitic stainless steels. Tap water, 3% NaCl, distilled water, humid air, and acetic acid (HAc) with or without "poisons" such as H_2S are among the more common corrodents. Solutions of NaCl containing HAc and H_2S have been used extensively in connection with cracking tests of candidate steels for the oil industry; data presented below would seem to indicate this to be a particularly vicious cracking medium.

A complete description of the result of a stress-corrosion experiment would require a statement of the rate of nucleation and the rate of growth of cracks as a function of stress concentration. This is a formidable task because, among other things, the nucleation and growth of one crack decreases the stress in adjoining areas, which has the effect of tending to inhibit the nucleation and growth of cracks at nearby sites. It is almost universally the practice to report only a single parameter of

cracking, namely, the total time under exposure until a complete rupture occurs. This of course gives a minimum amount of information about the intensity of attack, and it does not even distinguish between the "incubation" time and the growth time of the crack which terminates the test. In the case of self-stressed bent-beam specimens, the growth of cracks tends to decrease the stress elsewhere in the specimen to the extent that the nucleation and growth of a number of partial cracks may diminish the stresses to the point that no crack propagates clear across the specimen.

In spite of these obvious deficiencies of the "time-to-failure" parameter, the great expense of instrumenting specimens for a more thorough description of cracking kinetics makes it unlikely that another measure will supplant the customary one, at least for wide-scale tests.

SUMMARY OF FINDINGS

Mechanisms

The first procedure used in attempting to distinguish SCC from HEC was based on the position that the term SCC would be reserved for cracking such that corrosion occurs at anodic areas at the advancing tip of the crack, that small electric currents flow between local anodes and local cathodes to effect this corrosion, and that impressing small cathodic currents from an auxiliary electrode should tend to polarize the local cathode areas in the direction of the potential of the anodes and thereby should tend to mitigate the corrosion and the attendant cracking. Figure 8 shows the effect of small currents (note that these are of the order of microamperes per square centimeter) on the time to fracture for type 410 stainless steel under the indicated conditions. The fact that a low level of cathodic current very greatly extends the breaking time is taken to indicate that the cracking process in the absence of impressed current is properly classified as SCC. At high current densities, the specimens also fracture after a short time; this is taken to be HEC.

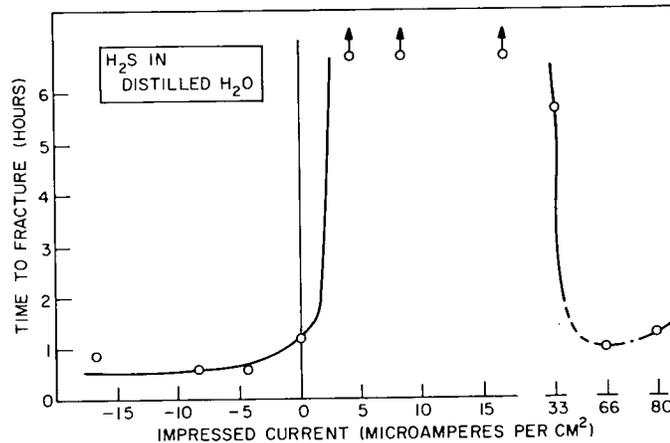


Fig. 8 - Effect of impressed currents on time to fracture of a martensitic steel (corrodent as noted)(9). A positive current in Figs. 8, 9, and 10 indicates the specimen is being made cathodic; a negative current indicates anodic.

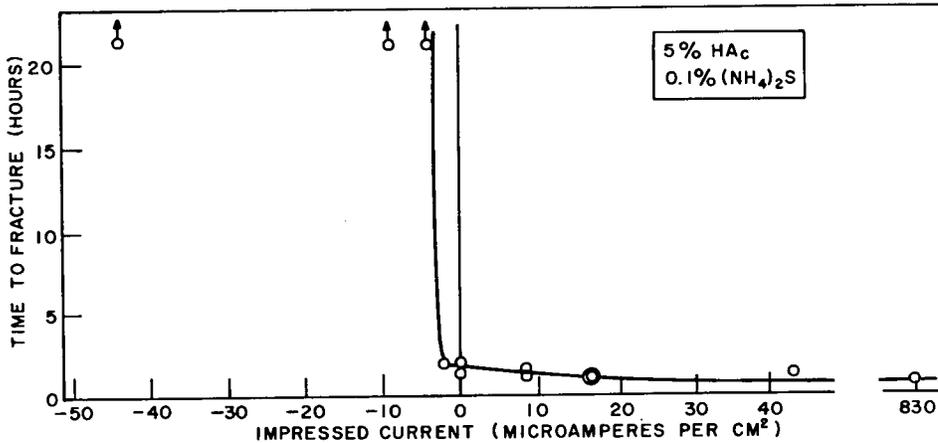


Fig. 9 - Effect of impressed currents on time to fracture of a martensitic steel (corrodent as noted) (9)

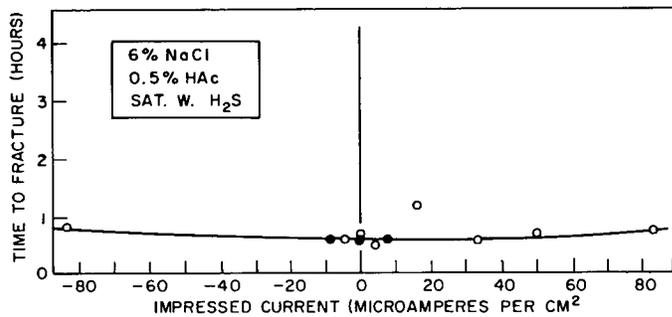


Fig. 10 - Effect of impressed currents on time to fracture of a martensitic steel (corrodent as noted) (9)

Figure 9 shows the results of a similar study but in a different medium. Here the imposition of even a small cathodic current decreases the breaking time, whereas the imposition of anodic currents greatly extended the life; this behavior is taken to indicate that in the case of zero impressed current (in the "natural" cracking case) the failure occurs by HEC.

Figure 10 shows a similar study except with the solution of NaCl containing HAc and H₂S mentioned before as a particularly vicious cracking medium. Here it is to be noted that impressing anodic or cathodic currents on the specimen was without effect on breaking time. One possible explanation is that the two processes can occur in overlapping ranges of current density.

These electrochemical analyses must be viewed as highly simplified and preliminary. They are, however, consistent with two subsequent observations: Conditions presumed by this analysis to cause SCC caused the measured potential to behave during cracking as if the cracking were occurring along an active path; and conditions presumed to cause HEC caused hydrogen to permeate a thin iron membrane.

The foregoing will be recognized as a highly incomplete and tentative statement of mechanisms, but the paucity of fundamental studies and the complexity of the phenomena leave little more than can be said at this stage.

Corrodents

Unlike SCC in austenitic stainless steels, which is believed to require either halides or caustic, cracking of the high-strength steels can be caused by a very large number of fundamentally different reagents, including humid argon. Mention might be made that the effectiveness of some of the very mild media was observed on specimens which had been coated with coatings commonly regarded as good corrosion barriers.

Susceptible Steels

All the high-strength steels, including maraging and precipitation hardening as well as martensitic, have been found to be susceptible to cracking. There are numerous indications that this susceptibility diminishes with decreasing strength in any steel, but no clear-cut threshold strength level for susceptibility has been delineated. The cracking behavior also diminishes with decreasing stress, but no sign of a clear-cut minimum stress for cracking has been found. Mention might be made, however, of the observations that thumbnail cracks such as in Fig. 1 grew rapidly when the steel was stressed beyond the yield in tap water, but not if stressed under the yield.

Metallography

SCC in aluminum alloys is consistently intergranular and in austenitic stainless steels is transgranular if caused by halides. When one tries to characterize the metallography of stress-corrosion cracks in the high-strength steels, however, he immediately faces the problem of deciding what to call a grain boundary. Even if he avoids or surmounts this problem, he may be no better off as far as arriving at a diagnostic criterion of cracking mechanism in the high-strength steels, at least based on limited experience with steels of the 4340 type. In these steels, cracks which are intergranular with respect to the prior austenite grains have been generated in the laboratory by stress corrosion, hydrogen embrittlement, impact, and thermal shock, and they are indistinguishable even by electron fractography, as far as the technology of today is concerned. The important point is that there is absolutely no grounds for concluding that a crack in a martensitic steel through a prior austenite grain boundary identifies the cracking process as stress corrosion.

CONCLUSIONS AND DISCUSSION

Any of the high-strength steels can be cracked by the right combination of heat treatment and environment. This includes many if not most of the promising commercial steels heat treated to the strength levels of particular interest to aircraft, rocket motor casings, and advanced deep-diving submarine hulls, and this cracking has been observed in environments which can hardly be classified as unrealistic or extreme.

These findings do not necessarily of course eliminate all or any of these materials from consideration. They simply mean that the effect of conditions of stress, environment, and heat treatment must be thoroughly explored in order to be able to live with the problem in these metals as it has been possible to live with the problem of SCC in aluminum alloys for aircraft.

The consequences of SCC to the structures of interest make it imperative to do a thorough job of assessing susceptibility. Until we understand SCC and HEC much better than at present, one should test a high-strength steel in any new environment contemplated regardless of inertness in other environments. Tests should always include specimens stressed above the yield strength regardless of the design stresses in the finished article: Even in structures of very simple geometry there may well be irregularities large enough to cause local yielding. Even if a coating is definitely specified for a given application, tests should be run on uncoated as well as coated specimens. This is partly because of the possibility of damaged coatings in practice and partly because of the experience that coatings on test specimens are consistently of better quality than on the production article.

The requirements of new steels for strength, toughness, and fabricability are so difficult to meet in the high-strength range that we cannot afford to reject a given steel because of an insufficiently sophisticated SCC test. For example, one can hardly afford to reject a given steel simply because a specimen broke in a SCC test if this fracture is really largely a brittle fracture perhaps triggered by a corrosion pit. The steel may be entirely suitable for reasonably long service in thinner sections. Also if in a given medium the cracking mechanism can be identified as SCC rather than HEC there is some possibility, though not a particularly attractive one, of using cathodic protection schemes for preventing cracking. Likewise, if cracking can be identified as HEC, anodic currents might afford a means for mitigating the cracking.

Prudence then dictates continued wide-scale testing, and economics recommend the development of very much more sophisticated testing than is now being used.

ACKNOWLEDGMENT

To Dr. S. W. Strauss the author is indebted for much of the task of searching the literature.

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U-bend strips of a martensitic stainless steel containing 12.5% Cr, 0.1% Ni, 0.12% Cu, and 1.2% Si were immersed in 3% NaCl solution. Rusting and pitting soon occurred, but cracking occurred only after 4-1/2 months. One specimen stressed as above was coupled to a sheet of Al foil and submerged in the salt solution. With this arrangement the specimen cracked overnight. When a Pt anode was used in a salt solution cell with the stressed specimens as cathodes, the specimens cracked. In a dilute H_2SO_4 solution containing a few drops of P dissolved in CS_2 to accelerate hydrogen adsorption by the alloy, cracking again occurred. Specimens were found to be very brittle after brief cathodic polarization in NaCl or H_2SO_4 but regained their ductility on standing in air. The results of these experiments were interpreted as indicating that this type of cracking should be described as HEC (hydrogen embrittlement cracking) as distinguished from SCC. (One reference.)

2. Fraser, J.P., and Treseder, R.S., "Cracking of High Strength Steels in Hydrogen Sulfide Solutions," *Corrosion* 8:342-350 (Oct. 1952).

The following steels were studied (figures in parentheses are the yield strengths in ksi): 9% Ni (106), 9% Cr (158), 12% Cr, wrought (102), 7% Cr (126), 5% Cr (97), SAE 4340 (126), N-80 (91), 5% Ni (61), J-55 (54), and 12% Cr, cast (55). Specimens were three-point-loaded beams with the stress beyond the yield of most of the alloys. Both H_2S and liquid water are necessary to crack these steels. H_2S in distilled water alone cracked only the 9% Ni steel; addition of 6% NaCl or of 0.5% acetic acid increased the number of steels cracked. The susceptibility of the steels to cracking varied roughly in proportion to their yield strength. Temperature had an appreciable effect on cracking tendency of borderline cases; for example one of two specimens of 5% Ni steel failed at 0°C in a solution of HAc plus H_2S , but none of three failed in the same solution at 25°C. Much of the cracking was reported to be transgranular and to be associated with pitting, which in turn appeared to be associated with inclusions. The authors think the cracking should be classified as SCC rather than HEC. (Seven references.)

3. Durkin, A.E., "Corrosion Cracking of Martensitic Stainless Steel," *Metal Progress* 64:72-75 (1953).

Strip specimens of type 410 stainless steel bent to produce various stress levels were exposed to (a) 10% NaOH, (b) cold concentrated HCl, (c) boiling concentrated HCl, and (d) 1:1 HCl plus SeO_2 . The susceptibility to cracking increased in the order (c), (a), (b), and (d). In a separate set of experiments, angle of bend to fracture was used to measure embrittlement. Immersion in corrodent (1:1 HCl plus SeO_2) before testing caused embrittlement in a subsequent bend test (dry); but an intervening bake restored ductility. From this it was thought that the static cracking was due to hydrogen embrittlement. (Two references.)

*The nomenclature of steels used in the abstracts follows the usage of the author of the original paper or report.

4. Badger, W.L., "Stress Corrosion of 12% Cr Stainless Steel," Trans. SAE 62:307-313 (1954).

Type 403 and three other 12% Cr steels (Mo-V, W-Ni, and Mo-W-V) were heat treated to several strength levels, and strip specimens were stressed as bent beams at four stress levels and immersed in 1:1 HCl plus 1%SeO₂ at room temperature. The resulting cracks resembled metallographically the service cracks observed in aircraft compressor rotor blades. There may be a minimum stress (roughly 50,000 psi) below which cracking does not occur. The four alloys are compared for cracking time at various hardnesses. It is noted that the frequency of cracking of compressor blades in service increases sharply with decreasing operating temperature, suggesting either that the cracking mechanism is hydrogen embrittlement or that notch sensitivity overrides the slow crack growth if this is by SCC. (Five discussions.)

5. Bloom, F.K., "Stress Corrosion Cracking of Hardenable Stainless Steels," Corrosion 11:351t-361t (1955).

U-bend and dead-loaded tension specimens were stressed at levels of 50 to 100 ksi and for the most part in a corroding medium consisting of distilled water containing 1/2% HAc saturated periodically with H₂S. (Other corrodents used were salt fog, salt solution, and marine and industrial atmospheres.) The following steels were tested at the Rockwell hardness levels indicated in parenthesis: 410 (C21-43); 416 (C40-42); 440C (C54-55); 440A (C52-53); 420 (C50-52); 17-7 PH (C44-46); 17-4 PH (C41-45); 431 (C42-43); 17-10 P (C30); SAE 4130 (C21-28); SAE 4340 (C36). For a given corrosion system, cracking was related to alloy hardness as well as to other parameters such as heat treatment and applied stress. A plot of time to failure versus hardness for specimens exposed to the HAc plus H₂S environment indicated that the average time to failure decreased with increase in hardness, reaching a minimum holding time, after which the holding time increased with further increase in hardness. The most severe media for promoting cracking in these steels, according to Bloom, were those which promoted hydrogen embrittlement. In the HAc plus H₂S tests, cracking occurred at hardness levels as low as Rockwell C24, in severe marine atmospheres at C40 and above, in severe industrial atmospheres at C45; salt fog and aqueous salt solutions caused cracking at Rockwell C45 and higher. The precipitation hardening alloys, 17-4 PH and 17-7 PH, were found to be susceptible to SCC if aged to their maximum hardness but were among the most resistant materials if overaged. Severely cold-worked types 304 and 305 cracked only in an acid plus H₂S medium containing NaCl; the same was true of 17-10 P hardened to Rockwell C30. Tempering at 1000° to 1050°F prevented cracking of 410 stainless steel in the severest media used. It was suggested that cracking is propagated by a combined mechanism which involves stress corrosion accelerated by hydrogen precipitated in deformed metal just ahead of the crack. (Eleven references.)

6. Fontana, M.G., "Stress Corrosion Cracking in Type 403 Stainless Steel," WADC Tech. Report 56-242, 1956, 51 pp.

Three-point-loaded beam specimens of type 403, 420, and 431 stainless steel stressed at levels 50 to 75 ksi were exposed at room temperature to a 1:1 aqueous HCl solution containing 1% SeO₂. The results indicated that the mechanism of cracking is the same for these steels. Cracking was associated with pits initiated at MnS inclusions, the pits propagating in a direction normal to the tensile stresses. Cracking did not occur in type 403 when it was tempered at 1050°F or higher, nor in type 420 tempered at 900°F or higher, nor in type 431 tempered at 700°F or higher. Electron microscopy and diffraction studies made on tempered type 403 indicated several carbides to be present after tempering in the 1000° to 1200°F range. A study was made of the effect of austenitizing temperature on hardness, impact strength, and microstructure of these steels.

7. Lillys, P., and Nehrenberg, A.E., "Effect of Tempering Temperature on Stress-Corrosion Cracking and Hydrogen Embrittlement of Martensitic Stainless Steels," Trans ASM 48:327-355 (1956).

Three-point-loaded beam specimens of type 410 (hardnesses R_c 28-40), 420 (R_c 31-50), 422 (R_c 33-47), and 436 (R_c 34-46) were set at a stress of 70 ksi for the type 410 and at 100 ksi for the other three steels. These were exposed in a salt spray cabinet; similar tests were made in a 0.1 normal solution of sulphuric acid containing arsenic, with the specimens being continuously cathodically charged. The susceptibility of 410 to cracking during cathodic charging, and the notch sensitivity (as determined by impact tests) followed similar trends as the tempering temperature was varied from 200° to 1000°F. Cracks in the 410 specimens cathodically charged were transgranular for specimens tempered below 850°F and were mixed transgranular plus intergranular for specimens tempered at higher temperatures. No cracking was observed in the salt spray test for this steel tempered at temperatures of 600°F or lower. In the salt spray tests the specimens in general fractured completely, shortly after the appearance of the first crack. (Twenty-two references; discussion.)

8. Brown, B.F., "Cracking of Martensitic, Type 410, Stainless Steel in Corrosive Environments," Report of NRL Progress, pp. 40-42, May 1958.

Increasingly larger polarizing currents, both anodic and cathodic, were applied to electropolished wire specimens which had been tempered at 900°F, stressed at 144 ksi, and immersed in (a) 5% HAC solution containing 0.1% $(NH_4)_2S$ and (b) a solution of H_2S in distilled water. The results showed that in (a), small cathodic currents increased the time to failure and anodic currents as well as higher cathodic currents decreased the time to failure, and in (b), cathodic currents decreased the time to failure and anodic currents increased the time to failure. These results were interpreted as indicating that 410 can fail by conventional SCC in (a) in the absence of cathodic polarization or in the presence of anodic polarization, and that with a sufficiently high cathodic current the steel begins to fail by hydrogen embrittlement; in (b) the fracture without any impressed current occurs by hydrogen embrittlement.

9. Halbig, J., and Ellis, O.B., "Observations on Corrosion Resistance of High Strength Stainless Steels for Aircraft," Corrosion 14:389t-395t (1958).

Only a small portion of this report deals with SCC. The authors mention briefly the work of F.K. Bloom, and then they present the following additional information on SCC: (a) in a mild industrial atmosphere, both 17-7 PH and PH15-7Mo in conditions TH1050, RH950, and CH900 have shown complete immunity to SCC when exposed for 21 months even when the specimens were stressed at 90% of their yield strengths, (b) both 17-7 PH and PH15-7Mo, in conditions TH1050 and CH900, did not crack at 50% yield stress (approximately 100 ksi) or at 75% yield stress after a 21-month exposure in a marine atmosphere, however, in condition RH950, failures occurred at 50% yield stress, but the data were not complete enough to permit comparisons to be made between 17-7 PH and PH15-7Mo. The authors state that stress corrosion failures of PH15-7Mo did not result in complete fracture of the specimens, but that failures of 17-7 PH generally did result in complete fracture of the specimen. (Twenty-six references.)

10. Brown, B.F., "Notch Sensitivity Effects in Stress Corrosion and Hydrogen Embrittlement Tests on High Strength Steels," Corrosion 15:399t-402t (1959).

Strip specimens of modified type 422 steel tempered at 875° and 1075°F were stressed in direct tension and exposed to a 1% HAC solution in water saturated with H_2S and NaCl. The specimens fractured partly by SCC and partly by terminal fast (purely mechanical) fracture. The extent to which the SCC must grow is determined among other things by the notch sensitivity of the steel. Highly brittle steels fracture with only a

small SC crack; failure to distinguish between these two fracture modes may give an erroneous impression of the rate of SCC, (Four references.)

11. Lochen, R.E., and Miller, E.R., "Stress Corrosion of 12% Chromium Stainless Steel," *Ind. Eng. Chem.* 51:763-764 (1959).

Flat, surface-ground test specimens of type 403 (with 0.05 Mo added), 403 (with 0.50 Mo added), and 422 stainless steel of Brinell hardness 205-372, 230-372, and 276-509, respectively, were stressed to 10 to 100 ksi by bending in a constant load fixture and exposed for 500 hours or until failure at 80°F to a 0.5% HAc solution saturated with hydrogen sulfide. Time to failure is shown as a function of stress. The maximum susceptibility to SCC (steel unspecified) corresponded to a tempering temperature of about 800°F, and the susceptibility in general was noted to parallel the notch sensitivity as measured in the Charpy impact test. Cracks were predominantly intergranular. (Fourteen references.)

12. Naumann, F.K., and Carius, W., "The Importance of the Corrosion Processes in Aqueous Hydrogen Sulfide Solutions for the Formation of Fracture in Steels," *Arch. Eisenhüttenw.* 30:283-292 (1959).

Specimens of steels of Vickers hardness 377-612 were stressed in bending, exposed to H₂S solutions at various potentials, and the time for fracture noted. In the cathodic reaction, atomic hydrogen is said to have been produced not by discharge of ions but rather by reduction of dissolved H₂S, and this process apparently made possible a high corrosion velocity which resulted in fracture of the stressed samples. It was concluded that the cracking mechanism was hydrogen embrittlement rather than SCC. (Nineteen references.)

13. Srawley, J.E., "Hydrogen Embrittlement Susceptibility of Some Steels and Nonferrous Alloys," *NRL Report 5392*, Oct. 1959, 25 pp.

Notched round bars of 4340 (R_c 26-52), 4130 (R_c 21-51), 422-M (R_c 42-51), 410 (R_c 21-41), and three precipitation hardening steels (R_c 46, 40, and 42) were cathodically charged and statically loaded (in air). HEC was observed in 4340 at hardnesses as low as R_c 26, in 4130 at R_c 21, in 422-M at R_c 42, in 410 at R_c 21, and in the three precipitation hardening steels. (Twelve references.)

14. Phelps, E.H., and Loginow, A.W., "Stress Corrosion of Steels for Aircraft and Missiles," *Corrosion* 16:325t-335t (1960).

Bent beam specimens of X-200, USS Strux, MBMC No. 1, 4340, 4130, hot-work die steel, 12 MoV, Stainless W. precipitation hardening steel (2 compositions), 17-5 MnV stainless, Tenelon stainless, 201, 202, and 301 were exposed in a marine atmosphere (Kure Beach). Similar specimens of X-200, USS Strux, hot-work die steel, 12 MoV, Stainless W, and a precipitation hardening stainless steel were exposed to the atmosphere at Monroeville, Pa. Cracking behaviors are given for these tests. In both groups the stress level in the specimens was set at 75% of the yield strength. "Very rapid" propagation of the cracks was noted in some of the 0.050-in.-thick specimens of 12 MoV stainless, but was not noted with 0.020-in.-thick specimens. (This is possibly attributable to the consequences of thickness to the size of flaw required to set off running brittle fracture.) U-bend specimens of 12 MoV immersed in 3% NaCl and subjected to various levels of anodic and cathodic currents cracked in a way which suggests that the cracking in the absence of impressed currents is by SCC rather than by HEC. Cracking under anodic conditions appears to be associated with a greater multiplicity of crack initiation than under cathodic charging. Atmospheric tests of the effect of coatings on two steels showed that Ni-Cd electroplate, aluminum, and zinc-dust tributyl titanate (all of which are anodic) and a rust-preventive wax were completely effective in preventing SCC, and ten other coatings delayed it. (Eight references; discussion.)

15. Scharfstein, L.R., and Eisenbrown, C.M., "Potential-Time Curves Obtained During the Stress Cracking of Metals," *Nature* 188:572-573 (1960).

Two 12% Cr steels stressed at 80% of their yield strength were immersed in (a) a solution of 1% SeO₂ in 1% HCl and (b) a solution of 1% SeO₂ in 5% H₂SO₄, and potential-time curves were recorded. Various currents were impressed on the specimens, and the time to rupture was recorded. The following observations were made: There is a gradual drop in potential followed by a slow rise and, at rupture, by a sudden rise to a more noble potential. This is explained tentatively as a consequence first of the reduction of selenium and then fracturing of the steel by HE. In a solution of 0.1% ammonium sulphide in 5% HAc the steels also crack by HE, but the potential does not rise therewith. This is tentatively explained as due to the absence of an insulating film, the reduction of hydrogen being the only cathodic process. (Four references.)

16. Suss, H., "Chromium Electroplates for Corrosion Protection of Stressed AISI 410 Steel in High Temperature, High Purity Water," *Corrosion* 16:497t-502t (1960).

Autoclave tests on unplated bent beam specimens tempered at 650°F and exposed to high purity air-saturated water at 300°F failed in 2 to 8 weeks. No failures occurred in hydrogen-ammoniated or lithium-hydroxide-treated waters in 6 months. Chromium-plated specimens gave anomalous results from which it was concluded that such plating should not be used for these environments and this steel, which behaved satisfactorily if tempered at 1125°F. (Seven references. See also *Metal Progress* 82(11):89-93 (1962).)

17. Trozzo, P.S., and McCartney, R.F., "Relationship of Microstructure and Stress Corrosion Cracking of Type 410 Stainless Steel," *Corrosion* 16:26-30 (March 1960).

Three heats of this steel that exhibited different degrees of susceptibility to SCC when exposed in the stressed condition (with an applied stress of 40-80 ksi) to 300°F air-saturated water in static autoclaves were investigated by electron microscopy and electron diffraction in order to determine the relationship between microstructure and SCC. These specimens, of hardnesses R_c 39 to 42, tensile strengths 182-192 ksi, and yield strengths 139-142 ksi, had received the following treatment in the unstressed condition: (a) placed in a furnace at 600°F, (b) temperature increased at the rate of 400°F per hour to 1500°F, (c) soaked at 1500°F for 1 hour, (d) temperature increased rapidly to 1800°F and held for 20 minutes, (e) inserted into a salt bath at 1050 to 1075°F and held for 8 to 10 minutes, (f) air cooled to room temperature, and (g) immediately tempered for 2 hours at 650°F and air cooled. It was found that the specimens cracked at stresses as low as 40 ksi along networks of essentially untempered martensite in the prior austenite grain boundaries. Tempering at a temperature of 1175°F, which was high enough to break up the continuous network in the grain boundaries did reduce the susceptibility to SCC but at the expense of a reduction of tensile properties. (Five references.)

18. Ault, R.T., "Stress Corrosion of Notched and Unnotched AM-350 Alloy," WADD Tech. Note 60-95 (1960), 11 pp.

Notched ($K_t = 4.57$) and unnotched strip specimens heat treated to a yield strength of 180,000 psi were stressed in tension and exposed to alternate immersion in 3-1/2% NaCl at room temperature. The stress in the smooth specimens was 60% T_y , and the load on the notched specimens was adjusted to give a calculated stress of 60% T_y at the notch root. All the unnotched failed within 150 hours, but none of the notched had failed in 800 hours; this is attributed to the smallness of the zone at the root of the notch at a stress equal to the stress everywhere in the gage length of the smooth specimens, indicating that some sort of singularity nucleates the crack. Premature fracturing was sometimes observed at crevices at attachments, attributed to crevice corrosion. (Three references.)

19. Bhatt, H.J., and Phelps, E.H., "Effect of Solution pH on the Mechanism of Stress Corrosion Cracking of a Martensitic Stainless Steel," *Corrosion* 17:430t-434t (1961).

Bent beam specimens of 12 MoV steel at 203,000 psi yield strength were stressed at 75% of this value and immersed in 3% NaCl solution saturated with oxygen. The pH of the solution was adjusted with HCl or NaOH. The corrosion rate was greatest and time to fracture was shortest at the lowest pH studied (pH = 1), and the time to failure gradually increased while the corrosion gradually decreased as the pH was increased from about 3 to 11. At pH values greater than 11, SCC did not occur and corrosion rates were found to be negligible. Results of polarization studies were, in general, similar to those obtained by the authors of items 5 and 22 in this bibliography. Hydrogen diffusion experiments were performed, and the results indicate that without applied current at pH₁ or with cathodic polarization at relatively high current densities at pH values of 1, 6.5, and 12.5 hydrogen entered and diffused through the steel. These results were interpreted as indicating that hydrogen enters and diffuses through 12 MoV stainless steel when this steel is cathodically polarized at current densities high enough to cause cracking, and that the cracking is due to hydrogen embrittlement. (Three references.)

20. Johnson, H.H., "Environmental Reactions of Rocket Steels," Cornell University, Progress Report No. 1, NRL Tech. Memo 155 (NRL Project 62R05-19A, Contract Nonr-3286 (00)(X)), April 1961, 10 pp.

Sheet specimens of AM 355 heat treated to 210 ksi yield strength and X-200 heat treated to 230 ksi yield strength were sharply notched (root radius approximately 0.001 in.) and broken under cyclic loading in air, water, or argon saturated with water. The X-200 was observed to be highly sensitive to water and to water vapor, both of which caused accelerated crack growth rate under fatigue loading. The cracking behavior of AM 355 on the other hand appeared to be little affected by water or water vapor.

21. Johnson, H.H., "Laboratory Evaluation of High Strength Steels," Cornell University, NRL Tech. Memo. 17 (NRL Project 62R05-19A), Oct. 1961, 2 pp.

A brief discussion is presented of environmental parameters which may affect the susceptibility of high-strength steels to fail prematurely. Water, either in the liquid form or as water vapor is said to be the most damaging environment, the deleterious influence of water becoming more pronounced as the steel strength is increased above 200 ksi. Crack growth which precedes catastrophic failure is said to be controlled by the amount of water vapor (relative humidity) in the environment. (Four references.)

22. Saibel, E., "Failure of a Thick-Walled Pressure Vessel," *Ind. Eng. Chem.* 53:56A-57A (July 1961).

A pressure vessel of 4-1/2-inch wall thickness was made from 4340 steel having a yield strength of 126 ksi. The pressure vessel was operated at room temperature with a maximum wall stress of 58.5 ksi. Cycling was slow, about 4 or 5 cycles a day. After 211 cycles the pressure vessel failed catastrophically. (The pressurizing fluid was not given.) The fracture origin was an intergranular zone assumed to be SCC which triggered the failure.

23. Phelps, E.H., and Mears, R.B., "The Effect of Composition and Structure of Stainless Steels upon Resistance to Stress-Corrosion Cracking," First International Congress on Metallic Corrosion, London: Butterworths, pp. 319-327, 1962.

U-bend specimens of the following steels were studied (yield strengths shown in parentheses): Type 410 (148 ksi); 12 MoV (207 and 193 ksi); 302 annealed (37 ksi), and cold worked (198 and 247 ksi); and 304 annealed (44 ksi), and cold worked (120, 161, and 186 ksi). Tests on the 12 MoV in various 1-molar solutions showed the following: NaI

and NaBr cracked the steel sooner than NaF and NaCl. Phosphate, sulphate, nitrate, sulphite, perchlorate, and acetate sodium salts also cracked this steel at pH levels varying among the various salts from 3.6 to 12.3. Sodium nitrite, bicarbonate, carbonate, cyanide, and hydroxide did not cause cracking. The full series of steels were tested in aerated 3% NaCl 10% NaCl plus H_2O_2 10% $FeCl_3$ boiling $MgCl_2$ 3% NaCl plus 0.5% HAC, 3% NaCl plus H_2S , and 3% NaCl plus 0.5% HAC plus H_2S ; the results of these tests are tabulated. Extensive metallography of cracked specimens is reported. Generally the 12 MoV had branching cracks in all solutions except those containing H_2S .

Anodic and cathodic polarization experiments were conducted following the method of Ref. 8; the cracking of 302 in 3% NaCl plus H_2S was indicated to be SCC rather than HEC. (Two references.)

24. Robinson, R.B., and Uzdarwin, R.J., "Investigation of Stress-Corrosion Cracking of High-Strength Alloys," Aerojet-General Corporation, Annual Summary Report No. 2092 (Contract DA-04-495-ORD-3069) (1961), 13 pp. plus 28 tables and 25 figures.

U-bend specimens and bent beam specimens stressed to 75% of the yield strength were exposed (the test period ranging from 21 to 28 days) in environments representative of those to which the alloys are normally exposed during manufacture, hydrostatic testing, and storage of solid rocket motor cases. The test environments included air, high humidity atmosphere, distilled water, tap water, $Na_2Cr_2O_7$ inhibited water, an aqueous nitrate plus nitride salt solution, a rust inhibiting oil, a chlorinated degreasing solvent, and an aqueous soluble oil solution. The following alloys were used in this study (with yield strengths shown in parentheses): (a) B120VCA (137.1 to 166.2); (b) PH15-7Mo (199.4 to 238.6); (c) AM355 (199.1 to 302.0); (d) D6AC (190.6 to 253.9); (e) 300M (195.2 to 234.3); and (f) Vascojet 1000 (189.8 to 244.2). The specimens, of varying thicknesses, were held in type 304 stainless fixtures. The results indicate that (1) the most corrosive environments were distilled water, tap water, salt water, and high-humidity atmosphere, and that distilled water was apparently more aggressive than any of the other environments used, (2) alloys (e) and (f) were found to be more susceptible (in increasing order of susceptibility) to SCC than the other three alloys investigated, and (3) none of the six alloys tested were found to be susceptible to SCC in the chlorinated degreasing solvent, in the rust inhibiting oil, and in the $Na_2Cr_2O_7$ inhibited water.

25. Slunder, C.J., and Boyd, W.K., "Environmental and Metallurgical Factors of Stress Corrosion Cracking in High Strength Steels," Battelle Memorial Institute, DMIC Report 151, 1961, 21 pp.

This is a review article in which characteristics of SCC (as well as other types of cracking) are discussed. A number of theories of SCC are reviewed. Most of this report does not deal with SCC of high-strength steels. Reference is made to some current extensive programs concerned with studies on SCC of high-strength steel. (Thirty-three references.)

26. Slunder, C.J., "Stress-Corrosion Cracking of High-Strength Stainless Steels in Atmospheric Environments," Battelle Memorial Institute, DMIC Report 158, 1961, 38 pp.

This report presents a compilation of data accumulated on the SCC of bent beam specimens stressed below the yield point and exposed to laboratory and atmospheric environments. Information concerning heat treatments, mechanical properties, and test procedures is also given. The data were assembled from tests conducted by a number of industrial firms. A total of twelve stainless steels of the following types were investigated: (a) martensitic (USS 12 MoV), (b) martensitic precipitation hardenable (17-4 PH, and Stainless W), (c) semiaustenitic precipitation hardenable (17-7 PH, PH15-7Mo, AM350, AM 355), and (d) cold-rolled austenitic (301, 201, 202, USS TENELON, USS 17-5 MnV).

Test results did not permit quantitative comparisons to be made of the relative merits of the alloys because insufficient data had been accumulated, and much of the available information was considered preliminary since the tests were still in progress at the time this DMIC report was prepared. (Six references.)

27. Jackson, R.P., "Stress-Corrosion Cracking of 17-4 PH Stainless Steel," E.I. du Pont de Nemours and Co., Savannah River Laboratory Report DP-769, 1962, 7 pp.

Notched and unnotched four-point-loaded beam specimens were exposed in autoclaves to heavy water at $pD = 10.5$ and 464°F . In these tests eight sets of stressed specimens and one set of unstressed specimens (three specimens per set) were exposed concurrently to both the vapor phase and liquid phase in the same autoclave. Each of the nine sets of specimens received one of the following treatments prior to testing: (a) 4-hour heat treatment at 1100°F , air cooled, with the heat treatment and air cooling repeated using notched and unnotched specimens, (b) notched and unnotched specimens cold worked 48% followed by two 4-hour treatments as in (a), (c) unnotched specimens aged 4 hours at 1100°F , air cooled, cold worked 20%, annealed at 600°F for 24 hours, and air cooled, (d) notched and unnotched specimens aged 1 hour at 900°F , and air cooled, (e) notched specimens cold worked 48%, aged at 900°F for 1 hour, and air cooled, and (f) unstressed notched specimens treated as in (a). All specimens were stressed beyond the yield. After 3900 hours of exposure, three notched specimens in condition (d) failed in the vapor phase in 1072 and 3900 hours, and the notched and unnotched specimens in condition (e) failed in the liquid as well as vapor phase in less than 100 hours. The specimens in conditions (a), (b), (c), and (f) survived the 3900-hour test. Some pitting was observed in the specimens which had been cold worked before being aged at 1100°F but not in any other specimens.

28. Johnson, H.H., "Preliminary Comments on the Role of Water Vapor in Slow Crack Growth in High Strength Steels," Cornell University, NRL Tech. Memo 208 (NRL Project 62R05-19A), Apr 1962, 3 pp.

A center-notch specimen of H-11 steel heat treated to a yield strength of 230 ksi was stressed under (constant tensile) loading with provision for controlling the relative humidity (RH) at either 18% or 100%. The length of the slowly growing crack was monitored by an electrical resistance method. The crack growth rate increased appreciably when the RH was increased from 18% to 100% and decreased when the RH was decreased to 18%.

29. Johnson, H.H., "Calibration of the Electrical Resistance Method of Studying Crack Propagation," Cornell University, NRL Tech. Memo. 205 (Contract Nonr-3286 (01) (X)) (NRL Project 62R05-19A), May 1962, 10 pp.

The method employs a Kelvin bridge circuit and sends 20 amperes through the sample intermittently. Center-notched specimens of H-11 (230 ksi yield strength), X-200 (230 ksi yield strength), and AM 350 (210 ksi yield strength) were used to calibrate the system by measuring the surface trace of the crack (occasionally correcting for submerged contour by polishing down to the centerline).

30. Owen, C.J., "Stress Corrosion of High Strength Steels and Alloys; Artificial Environments," Mellon Institute, Final Report (Contract No. DA-36-034-ORD-3277RD), 1962, 110 pp.

Bent beam specimens stressed to 75% of the yield strength and U-bend specimens stressed to levels between the yield and the fracture strength were totally immersed in room temperature aerated 1-molar solutions of NaCl , NaNO_3 , Na_2SO_4 , NaPO_3 , and Na_2S . The following steels were studied: D6Ac, 300M, Vascojet 1000, AM 355, PH 15-7 Mo, 4137 Co, Rocoloy 270, and Ardeform 301. Titanium alloy B120VCA was also studied. The steels were tempered at various temperatures over a wide temperature range. The

relative cracking tendencies of the various materials are reported (but this relative ranking must be accepted with reservation unless it can be verified that all specimens were of the same thickness). The author apparently presumes that aeration precludes HEC.

31. Rowland, M.C., and Smith W.R., Sr., "Precipitation-Hardening Stainless Steels in Water Cooled Reactors," *Nuclear Eng.* 7:14-22 (Jan. 1962)

Stress corrosion tests in static and steam autoclaves and dynamic water and steam corrosion loops were made on the following steels having yield strengths indicated in parentheses: (a) 17-4 PH (150-185 ksi), (b) 17-7 PH (185-195 ksi), (c) AM350 (150-170 ksi), (d) A-286 (93 ksi), and (e) Stainless W. The test fixtures consisted of a series of 1/4-in.-diameter type 304 stainless steel pins pressed into precisely drilled holes in type 304 plates. The test specimen strips were bent around three pins to produce a deflection which had been calculated to give a maximum fiber stress of 30 ksi to stresses over the yield point. Examination of specimens were performed after 300 hours exposure and then at approximately 500-hour increments. The results of both the static autoclave tests and the loop tests showed that stressed specimens of 17-4 PH (in the condition H900) will crack in high-temperature water, and the failures are intergranular. Cracking was reported to occur rapidly after a certain incubation period. Most of the failed test pieces were completely fractured, although a few had only partial fractures. Aging at 1100°F to produce condition H1100 improved the resistance of 17-4 PH to cracking. No failures occurred at applied stresses slightly over the yield strength in 5400 hours of exposure. A comparison of 550°F test results obtained in the oxygenated loop with those obtained in the static autoclave tests at the same temperature indicates that oxygen even in small amounts (0.2 ppm) appears to promote cracking of 17-4 PH in the H900 condition. It was reported that cycling test temperatures between 200°F and 550°F resulted in more severe cracking than did exposure at any one temperature in this range. Static autoclave tests did not crack 17-7 PH aged at 1050°F, stressed to levels up to 120 ksi, and exposed for 2900 hours; specimens exposed in simulated Dresden reactor water survived 2800 hours at 550°F at an applied stress of approximately 280 ksi. Specimens of AM-350 and of A-286 survived a 4700-hour exposure in static autoclaves at applied stresses of 30 ksi to stresses over the yield point; the AM-350 specimens were also tested under cycling temperatures between 200° and 550°F, and survived a 4700-hour test. Some testing of A-286 was also done in both a high oxygen (1.2 ppm) loop and in a simulated boiling-water reactor (0.2 ppm oxygen) loop, and no failures occurred on specimens stressed to their yield strength; however, in the high oxygen loop, all specimens stressed considerably over the yield strength failed in a 700-hour exposure. (Seven references.)

32. Suss, H., "Shot Peening of Metals for Protection Against Stress Corrosion Cracking," *Corrosion* 18:17t-20t (1962).

This paper reviews studies which have been conducted on the effect of shot peening on the susceptibility to SCC of the martensitic stainless steels AISI 410 and AISI 4340. Shot peening has been found to protect hardened (RC36-42) AISI 410 stainless steel from SCC in 300°F high-purity water indefinitely at stresses up to approximately 45 ksi (about 1/3 of the yield strength) and for lesser but still useful finite periods at stresses up to 60 ksi. (Nine references.)

33. Weibull, I., "Stress Corrosion Cracking in High Strength Steel - Or Hydrogen Embrittlement," *Advances in Aeronautical Sciences*, Pergamon Press, 3:335-356 (1962).

Test pieces in the form of bolts put into steel bars were tightened by nuts, the tightening torque giving a nominal stress in the core section of the threaded part of approximately 75% of the yield strength, and exposed in a humidity cabinet holding air at 100% humidity at a temperature alternating between 25° and 43°C. Bolts stressed in the same way as described above but exposed to an ordinary indoor atmosphere did not fail during a 60-day

test period. The above failures were in the humidity tests and therefore attributed to the combined action of moisture and stress. In a second series of experiments specimens of three steels were loaded in tandem in a creep testing machine using an applied stress of 75% of the notch tensile strength. Tap water was led to some specimens by means of a wick of yarn, one end of which was dipped into water and the other wound around the notch of the specimen; when a specimen failed, it was replaced by a dummy. The results showed that nonwetted specimens had an appreciably longer lifetime than wetted specimens. Anodic and cathodic polarization studies were made in stressed specimens in distilled water from which it was concluded that the service cracks were probably caused by HEC rather than by SCC. (Five references.)

34. (Anonymous) "Oil Well Materials for Resistance to Hydrogen Sulfide or Stress Corrosion Cracking," *Materials Protection* 2(3):93-94 (Mar. 1963).

Report of Technical Unit Committee T-1B recommends tentative specifications which include tempering martensitic low alloy or stainless to a hardness of R_c 22 maximum.

35. Davis, R.A., "Stress-Corrosion Cracking Investigation of Two Low Alloy, High-Strength Steels," *Corrosion* 19:45t-55t (1963).

Strip specimens of 4330M (252.7 ksi ultimate) and 4340 (278.3 ksi ultimate) were bent as cantilever beams (with tapered cross sections to produce constant moment) and as 4-point-loaded beams. Experiments were made in 3% NaCl and in 0.25-normal Na_2S , with and without anodic and cathodic currents. No clear-cut microstructural differences were detected between fractures believed to be HEC and those believed to be SCC. Shot peening improved resistance to SCC. (Three references; discussions.)

36. Setterlund, R.B., "Investigation of Stress-Corrosion Cracking of High-Strength Alloys," Aerojet - General Corporation Report LO414-01-22, May 1963, 4 pp. plus 6 tables (Twenty-Second Monthly Progress Report on Contract DA-04-495-ORD-3069 with Frankford Arsenal).

Bent beam specimens of 18% Ni and 20% Ni maraging steels at yield strengths from 283 ksi to 354 ksi were exposed to distilled water, tap water, 3% NaCl, 0.25% sodium dichromate, air of two levels of humidity, and various other special organic environments. Also specimens of H-11 steel were used to evaluate the effectiveness of standard coatings in preventing cracking in moist air and in 3% NaCl. Both the maraging steels cracked in all levels of heat treatment in moist air, and most specimens failed in several of the other environments as well. All H-11 specimens surface-ground or sanded before being coated cracked in moist air regardless of coatings, which included polyurethane, zinc silicate, and six different epoxy coatings.