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MATHEMATICAL, PHYSICAL & ENGINEERING SCIENCES

Discussion

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2.7 The contributions of physical metallurgy and of fracture mechanics to containing the problem of stress-corrosion cracking

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[Plates 17 and 18]

The features in the province of the physical metallurgist – grains, precipitate particles, inclusions and lattice imperfections – have sizes approximately in the range of 1 μm–0.1 nm. The nature and properties of these microfeatures determine the engineering properties of alloys, including their resistance to stress-corrosion cracking. No basis has developed for confidence that knowledge of these microfeatures can be used to exclude the possibility of stress-corrosion cracking. Neither have we learned how to sum up the effects of these microfeatures to predict stress-corrosion cracking behaviour, nor is there any basis for predicting how soon we may be able to do this summing up analytically. In the meanwhile we need experimental means to do this summing up, and fracture mechanics appears to be an eminently suitable way to treat the stress factor. Other fundamentally different approaches may be found to be both valid and needed for special areas of the problem. Thus the role of fracture mechanics is to provide engineering data, and the role of physical metallurgy is to permit orderly alloy development through understanding of the data from fracture mechanics (and possibly other) tests. The two approaches are complementary, not competitive, and neither can fulfill the role of the other.

Non-redundant load path designs may require specially competent attention to both stress-corrosion cracking and brittle fracture.

INTRODUCTION

The invitation of the organizers of this conference to discuss the contributions of physical metallurgy and of fracture mechanics to stress-corrosion knowledge is especially welcome in view of what the author believes to be a widespread and prolonged misunderstanding of the roles of these two disciplines in the containment of the problem.

The practitioner of physical metallurgy contributes to the development and selection of alloys for engineering structures through his knowledge of the microstructure of alloy families and the contribution of various microstructural features to engineering performance. Microstructure is used here somewhat more broadly than usual and includes crystal structure, morphology of constituent grains, precipitates in the most general sense, including preprecipitation states of aggregation, texture, and the various types of crystal imperfections, including what Dr Rosenhain would possibly have termed mosaic structure. These features, having sizes spanning the range from about 1 μm–0.1 nm, determine the engineering properties of alloys.

Dr Rosenhain had the benefit of the art of the microscopy of metals due to Sorby's pioneering work a century ago, and while Rosenhain was still active X-ray diffraction had become a standard tool of the physical metallurgist. Since his day the number of experimental tools developed to aid the physical metallurgist has increased following the exponential pattern common to so much of science. During the 1950s, and following the lead of a contemporary of

Dr Rosenhain, A. A. Griffith, the methodology of (macroscopic) fracture mechanics was developed to the stage of engineering applicability, including eventually stress-corrosion cracking. The aim of this paper is to examine, in the light of all these developments, the contributions and roles of physical metallurgy and of fracture mechanics in controlling the problem of stress-corrosion cracking in present day engineering.

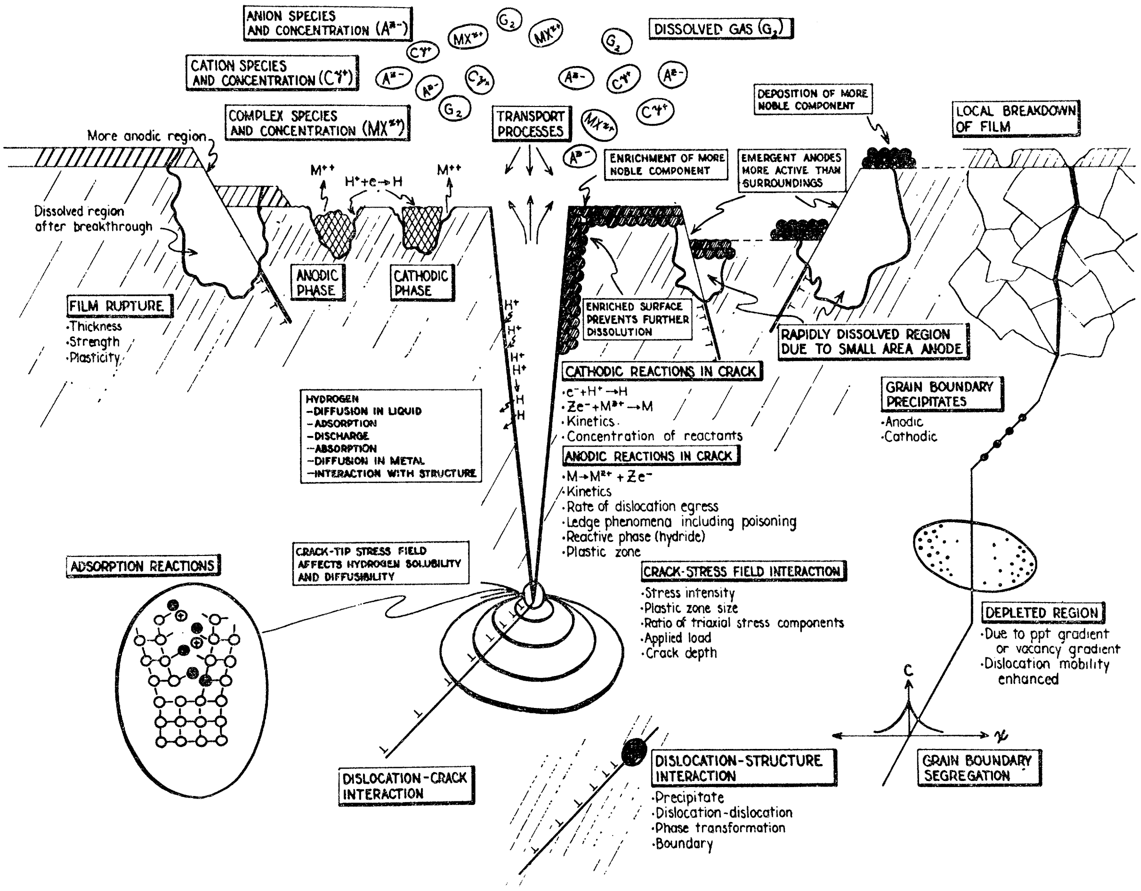

Figure 1. Schematic summary of microfeatures believed possibly involved in stress-corrosion cracking. Figure kindly supplied by Staehle (from Staehle 1969).

Most of the features and processes known or thought to be involved in stress-corrosion cracking in various systems are shown schematically in figure 1. There was a period when there appeared to be an unwritten creed in some circles that if the physical metallurgist or metal physicist became sufficiently clever in studying the 1 μm–0.1 nm features involved in cracking, measures could be devised to banish the problem, somewhat as the development of the vaccine banished the scourge of smallpox from much of the earth. Nothing has appeared on the horizon to support that outlook.

Since we have no reason for confidence that we will be free from the threat of stress-corrosion cracking but must somehow learn to cope with the effects of the microfeatures of figure 1, we must learn how to sum up those effects. We have absolutely no way to estimate how long it will take us to do this analytically. It could easily be the end of the century.

It is here that we can look for guidance from the progress which has been made in the closely related field of purely mechanical brittle fracture, the problem which caused the breaking apart of welded freighters in World War II, the bursting of the fuselages of the Comet III aeroplane, and numerous other catastrophes. It was known that the features familiar to the physical metallurgist are involved in brittle fracture – grain size, crystallography, precipitate particles, inclusions, and dislocations. The difficulty was that despite large and lengthy research programmes no one found a way to sum up analytically the contributions of these 1 μm–0.1 nm features to predict brittle fracture of engineering structures.

Fortunately, a second approach was made employing what might be called macro-mechanical metallurgy, to use R. F. Mehl's phrase, and including fracture mechanics in that term. The current technological bases for avoiding brittle fracture are the fruit of this macro-mechanical metallurgy approach.

These fracture control measures came about by the demonstration by Pellini and Puzak of a ductile-to-brittle transition temperature which, properly determined, affords positive protection against brittle fracture of ship plate and similar steel structures. We shall see below that an analogous approach may become useful in controlling stress-corrosion cracking in certain fields. For alloys which do not undergo a sharp transition, the fracture mechanics of Irwin and Kies affords a means of controlling the brittle fracture problem. Both the transition temperature method (standardized as ASTM E208) and the fracture mechanics method (standardized as ASTM E399) for controlling brittle fracture rely only on macroscopic specimens to sum up experimentally the effects of all the physical metallurgical microfeatures.

Through the use of these macroscopic methods we know quantitatively what changes in fracture toughness can be achieved by changing ferrite grain size, or by special vacuum melting, or by lowering the size of intermetallic compounds in high-strength steels. Turned the other way round, modern physical metallurgy enables us to understand the engineering properties of materials in terms of microfeatures, even though the properties must be measured macroscopically. Thus in brittle fracture the role of measurement is played by macro-mechanical metallurgy, and the role of physical metallurgy is to understand the reason for the measured properties in order to improve alloys in a non-random way. Physical metallurgy and macro-mechanical metallurgy are then not competitive approaches to the problem, but rather complementary approaches. Technological advances in the field require both.

The designer of submersibles used in the oil and gas fields in the North Sea specifies his hull steel by 'NDT + 60', meaning that the nil-ductility transition temperature must be at least 60 °F ($33\frac{1}{3}$ °C) lower than the minimum service temperature. A vast amount of research went into the development and validation of the 6-character specification inside the quotation marks above by which the designer sets his mind at ease regarding brittle fracture. The success in controlling the brittle fracture problem either by the transition temperature method or by fracture mechanics, as appropriate, has caused stress-corrosion cracking to replace brittle fracture as the most worrisome threat to structural integrity facing the designer, fabricator, and maintenance engineer in a wide range of technologies.

FRACTURE MECHANICS AND STRESS-CORROSION CRACKING

Returning to figure 1, in addition to the features of physical metallurgy we see ionic species, passivation, film rupture, and various other complexities of electrochemical corrosion. The inescapable conclusion is that if in the case of brittle fracture, where only physical metallurgy and mechanics are involved, we had to resort to a macroscopic specimen to sum up experimentally the contributions of all the microfeatures, there is surely even more reason to use the same approach when electrochemical corrosion has been added.

The macroscopic specimen approach and the physical metallurgy approach are not alternatives to solving the stress-corrosion problem. As with brittle fracture, they represent complementary approaches, one to produce characterization of engineering significance and the other to understand the reasons for the values measured.

But what sort of macroscopic test? After all, Roberts-Austen identified tensile stress as a necessary condition for this form of cracking using macroscopic specimens almost 90 years ago.

If one wishes to determine the stress at which a stress-corrosion crack will stop growing, or at which it will reinitiate, or if one is concerned with the effect of stress on cracking kinetics, it is necessary to quantify the stress effect in the presence of a crack. We do not have a valid way to do this except using fracture mechanics. This does not mean that to get stress-corrosion information of engineering utility one must use a fracture mechanics test. If I were asked to evaluate the effects of changing condensate composition on the cracking possibilities in Admiralty brass feedwater heater tubing, for example, I would probably prefer to cut a ring from the tubing, slot it to make a 'C-ring' specimen, and stress it as in service in the proposed environment. But the only reason for that preference would be that the material having the proper texture would be available for easy specimen manufacture. There is no reason that the same information could not be obtained using a fracture mechanics test, which would have the added advantage of being incorporable into a data bank of homogeneous information.

Three objections have been raised to the use of fracture mechanics to measure stress-corrosion characteristics: (1) The fracture mechanics specimen is provided with an initial crack-like flaw of controlled location and dimensions, and it has been the experience that designers, fabricators, and vendors may express umbrage at the implication that their product may be flawed. But the U.S. aerospace programme alone proved the wisdom of including an evaluation of the performance of materials containing a reasonable flaw. (2) Some of those experts who have been aware that the elastic fracture mechanics of Irwin and Kies is not useful to measure the brittle fracture characteristics of tough materials (because of the enormous size of specimens required to provide the necessary constraint to cause brittle fracture) have concluded that fracture mechanics is not therefore suitable for studying stress-corrosion cracking in the same alloys. But *all stress-corrosion cracks are brittle*, that is, they occur at stresses below that required for general yielding and therefore propagate in an elastic body even though there is *local* plasticity which is probably essential to the cracking process. Since the body as a whole is in the elastic range, elastic fracture mechanics is appropriately used to treat stress. (3) But the matter of local plasticity must be addressed. Indeed so, but the local plasticity is caused by and is therefore related to the surrounding macroscopic elastic field, even though we may not know the exact relationship. Thus the local plasticity is included with the other physical metallurgy features summed up by the macroscopic specimen, though one must bear in mind that room temperature creep is observed in *some* alloys stressed as precracked specimens.

Since the fracture mechanics approach is to observe the response of a cracked body to loading, experimentally one inserts a mechanical crack initially in order to obtain a crack of controlled location and geometry, commonly by machining a notch and fatiguing it. This procedure has special merit in assessing the stress-corrosion characteristics of weldments, especially for medium strength steels in seawater. In such materials stress-corrosion cracks usually initiate from corrosion pits, the location of which is determined by several electrochemical factors and may be unrelated to the relative vulnerability of the underlying metal to stress-corrosion cracking. If one works with a small sample of weldment, it is entirely possible that no pit forms at the locations which are most vulnerable to cracking, yet a large welded structure may have some pits at the sites of maximum vulnerability. This argument has given rise to the practice of siting precracks at the weld centreline, at the fusion line, in the heat-affected zone, and in the prime plate for stress-corrosion tests.

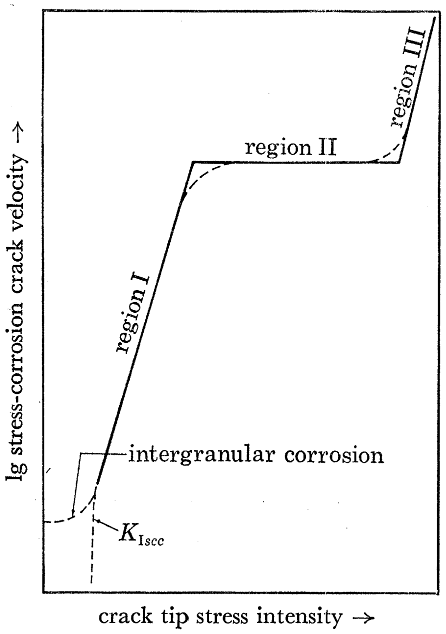

FIGURE 2. Schematic summary of the effect of stress-intensity K on stress-corrosion cracking kinetics. In some alloys region III is missing, and region II sometimes appears only as a point of inflexion. (From Speidel 1971.)

Having a cracked specimen, the elasticians have given us a theoretical method, and the continuum mechanicians an experimental method, to determine and express the stress intensity by a single parameter K. At sufficiently high values of K a stress-corrosion crack grows, and over the K range designated region I in figure 2, the cracking rate is approximately an exponential function of K. At higher K values there may be a plateau labelled region II in which some transport process is rate controlling and the cracking rate is independent of K. For some metals there may be a region III at the highest K levels in which cracking kinetics are again K-dependent. At sufficiently low K levels there may be a genuine threshold labelled $K_{\text{Is.s.c.}}$. Such a genuine threshold has been proven for titanium alloys, based on electrochemical measurements (Brown 1970). There are theoretical reasons to believe that such a genuine threshold exists for high-strength steels. But for aluminium alloys there is uncertainty about the existence of such a threshold (note figure 3).

The value of a number such as $K_{\text{Is.c.c.}}$ lies in its ability to predict the combinations of stress level, flaw size and shape which would be expected to produce stress-corrosion cracking through the Irwin equation

$$K^2 = \frac{1.21\pi\sigma^2 a}{\phi^2 - 0.2(\sigma/\sigma_y)^2},$$

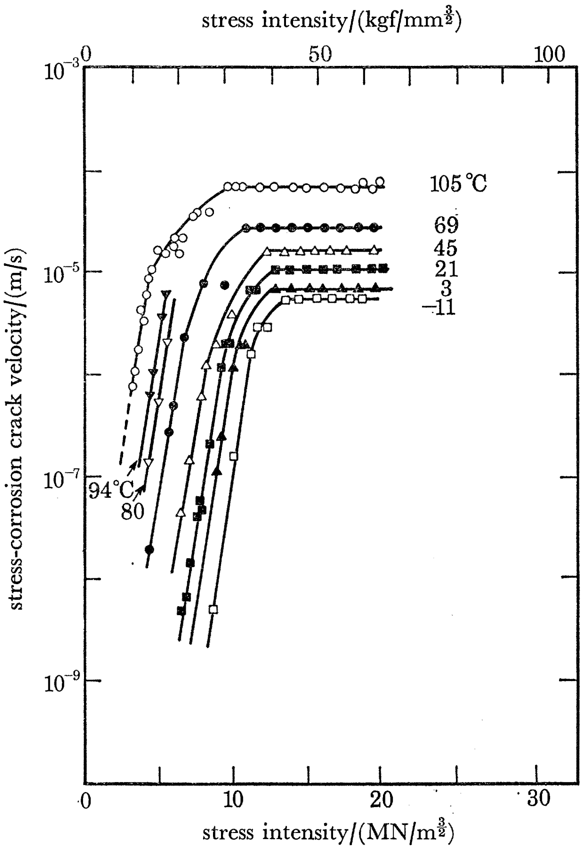

FIGURE 3. Stress-corrosion cracking kinetics in an aluminum alloy as a function of stress intensity at several temperatures. The activation energy in region I is much greater (27 kcal/mol; 113 kJ/mol) than in region II (3.8 kcal/mol; 15.9 kJ/mol), a fact which would not be suspected in the traditional undifferentiated total-time-to-failure data. (From Speidel 1971.)

Alloy 7079-T651; 2.5 cm thick plate; crack orientation TL; 3 m aqueous Kl solution; potential −450 mV *vs* E_{H_2/H^+}; pH 6.

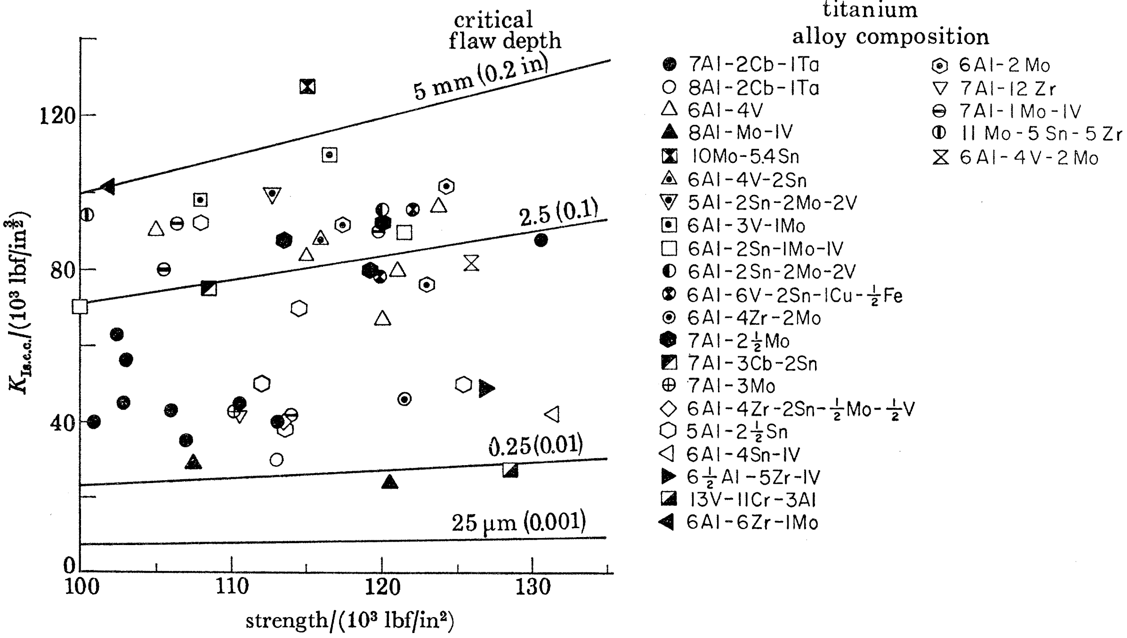

FIGURE 4. $K_{Is.c.c.}$ data for several titanium alloys in 3.5 % NaCl and seawater, plotted against yield strength. See text for explanation of critical flaw depth lines. Data of B. F. Brown, R. J. Goode, R. W. Judy, Jr, R. L. Newbegin and L. J. Waldron, U.S. Naval Research Laboratory. Figure kindly supplied by Judy.

where σ is the remote stress normal to the surface crack of depth a. σ_y is the yield strength, and ϕ is a shape factor. For $K = K_{\mathrm{Is.c.c.}}$, $\sigma = \sigma_y$, and for a crack at least 10 times as long as deep, the equation becomes

$$a_{\mathrm{cr}} = 0.2\ (K_{\mathrm{Is.c.c.}}/\sigma_y)^2.$$

$K_{\mathrm{Is.c.c.}}$ for a number of titanium alloys of varying yield strengths are shown in figure 4. In this figure are lines for four values of critical flaw depth, a_{cr}, with the simplifying assumptions noted immediately above. The significance of these lines may be illustrated as follows: If one has the same environment for which $K_{\mathrm{Is.c.c.}}$ was determined, and if he cannot be certain of inspecting out cracks as deep as 0.01 in (0.25 mm), he had better not use the two alloys having $K_{\mathrm{Is.c.c.}}$ values which lie below the 0.25 mm line. Another way to look at the data of figure 4 is that the load carrying capability of a member having a standard surface flaw is linearly proportional to $K_{\mathrm{Is.c.c.}}$.

Most of the data of figure 4 were taken using cantilever loaded bend bars. There have been two studies (Judy & Dahlberg 1968) in which by the use of the Irwin equation the predictability of cantilever beam $K_{\mathrm{Is.c.c}}$ data have been checked using large plates stressed in simple tension and provided with small surface flaws of known geometry. Both studies have confirmed the significance of the $K_{\mathrm{Is.c.c.}}$ number in panels bearing reasonable resemblance to large structures. There is a second kind of evidence of the validity of $K_{\mathrm{Is.c.c.}}$ data based on service failures. In the early days of the U.S. aerospace programme, rocket motor chambers of high-strength steel were observed to undergo stress-corrosion cracking during hydrostatic testing with tap water. Estimates of $K_{\mathrm{Is.c.c.}}$ indicated that a critical flaw depth for that steel in tap water would be about 25 μm. Because of incomplete control over heat-treating atmospheres in large furnaces in those days, there were numerous oxide intrusions at grain boundaries approximately 25 μm deep, and although the nominal hydrotest stress was somewhat less than the yield strength, in those days out-of-roundness was such as to elevate the stress locally to the yield level, undoubtedly cracking either alongside or through some of the oxide intrusions to provide the precrack. Thus all the criteria for stress-corrosion crack growth were met. To avoid the problem, the pressurizing fluid was changed to an inert oil. The conclusion from these experiments and experience, more meagre than I would prefer, is that if carefully measured fracture mechanics stress-corrosion data indicate a likely problem, it is a rash engineer who would disregard them in a critical circumstance.

It is nonetheless too early to know how fracture mechanics data on stress corrosion will fare in engineering design to contain the problem. There are reasons to believe that it may not be the best way to control the problem in certain technologies, such as steel pipeline, where the pioneering work of Parkins has shown that determining the electrochemical potential regime in which cracking occurs, and avoiding that regime, rather than measurement of cracking resistance itself in that regime, may be the way to go about controlling the problem. If so, this approach is exactly analogous to the use of transition temperature to avoid brittle fracture of ship plate.

Physical metallurgy

The engineering solution to cracking by certain environments, such as methanol or mercury around stressed titanium alloys, has been to simply prohibit the combination. In some circumstances, such as N_2O_4 in titanium bottles for space power plants, the engineering solution has been to add an inhibitor. But such solutions are not appropriate for aircraft or marine vehicles, for which the containment of the stress-corrosion problem had to be sought by the metallurgical route. There has been much progress in this way (Blackburn, Smyrl & Feeny 1972).

The phase α_2 (Ti_3Al) was found to promote aqueous stress-corrosion cracking, which explains why an alloy such as Ti–8%Al–1%Mo–1%V is so highly susceptible (see figure 4). Susceptibility in the titanium alloys containing aluminium starts at about 5% Al, but oxygen and tin contribute to the susceptibility. The 8–1–1 alloy mentioned, which consists of $\alpha+\alpha_2+\beta$, can be solution treated at about 1000 °C and quenched to produce α' or $\alpha+\alpha'$, with greatly increased resistance to cracking. But as the merits of various titanium alloys were better assessed, many of them such as the 8–1–1 alloy, Ti–7%Al–2%Nb–1%Ta, Ti–5%Al–5%Zr–5%Sn and a few others have become effectively obsolete. The alloy Ti–6%Al–4%V, preferably with oxygen less than 0.13%, and even more preferably less than 0.10%, has largely replaced the susceptible alloys in critical components for manned vehicles. New alloys of low susceptibility, based on present understanding, are expected to appear.

In the field of aluminium alloys, the texture of wrought alloys is a major tool in containing the stress-corrosion cracking problem. Not only is the textured alloy less susceptible than the equiaxed alloy, but additionally vulnerability is effectively restricted to one direction with respect to the texture, and design may be used to shield the vulnerable direction from sustained stress and corrosive environments. Manganese, chromium and zirconium in Al–Mg–Zn alloys are effective in pinning grain boundaries and thus retaining the texture after heat treatment. Iron and nickel are similarly effective in Al–Cu alloys. The research submersible Aluminaut was made of the Aluminum Association alloy 7079-T651, which is probably the most susceptible of any commercial alloy but which was nevertheless specified for its yield strength. From considerations of stress-corrosion cracking vulnerability, the vehicle was designed to be made of a series of rolled rings, with the vulnerable end-grain tucked under in each ring so as to be inside the hull rather than being washed by seawater. This vehicle had extensive service with no cracking problems. In view of Roberts-Austen's finding that tensile stress is necessary for stress-corrosion cracking to occur, one might reasonably ask why there should be any concern about the problem in the hull of a submerged hull, where the design stresses are compressive. The answer is that most stress-corrosion cracks in aluminium structures are due not to design stresses, but to stresses caused by heat treatment or assembly. In fact, one class of European submarines made of a special steel was scrapped because of the appearance of the stress-corrosion cracking problem due to assembly stresses. The principal metallurgical tool to avoid stress-corrosion cracking in Al–Mg–Zn alloys is overaging to produce the -T73 temper, even though this heat treatment entails some loss in strength.

There is a wide range of susceptibility of various heat-treated alloy steels in seawater. Strength is one factor which affects the degree of susceptibility, but even at a given strength level there is a wide range of susceptibility. The important work at the Welding Institute in Abington has shown that if twinned martensite is present, the $K_{\mathrm{Is.c.c.}}$ is low, and the crack path is along the prior austenite grain boundaries. In less susceptible steels, meaning that cracking occurs only at higher K levels, the crack path is transgranular, often called quasi-cleavage; the fewer large intermetallic particles present, the more resistant is the steel. Hence precipitation hardening steels and maraging steels can be made which are more resistant than tempered martensitic steels, meaning that cracks grow only at still higher K levels, and the fracture mode is one of microvoid growth.

The foregoing is cited as indicative of the body of knowledge, here greatly condensed, which

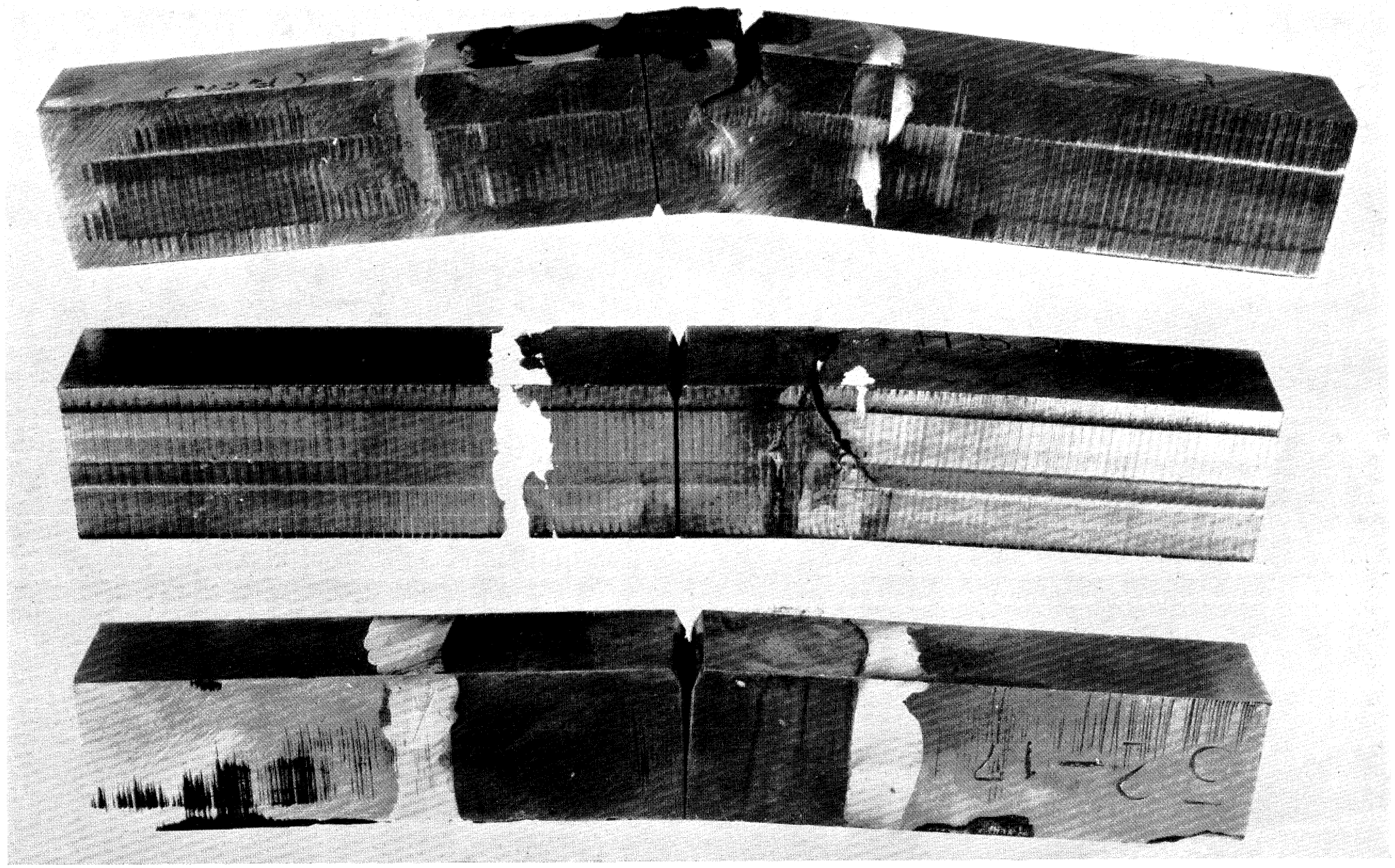

Figure 5. Specimens of the precipitation hardening steel PH 13-8 Mo after being tested. See text for details. Experiments demonstrate that electrochemical factors can override mechanical factors in selecting site of initiation of stress-corrosion cracks. Photograph courtesy of R. L. Newbegin, U.S. Naval Research Laboratory.

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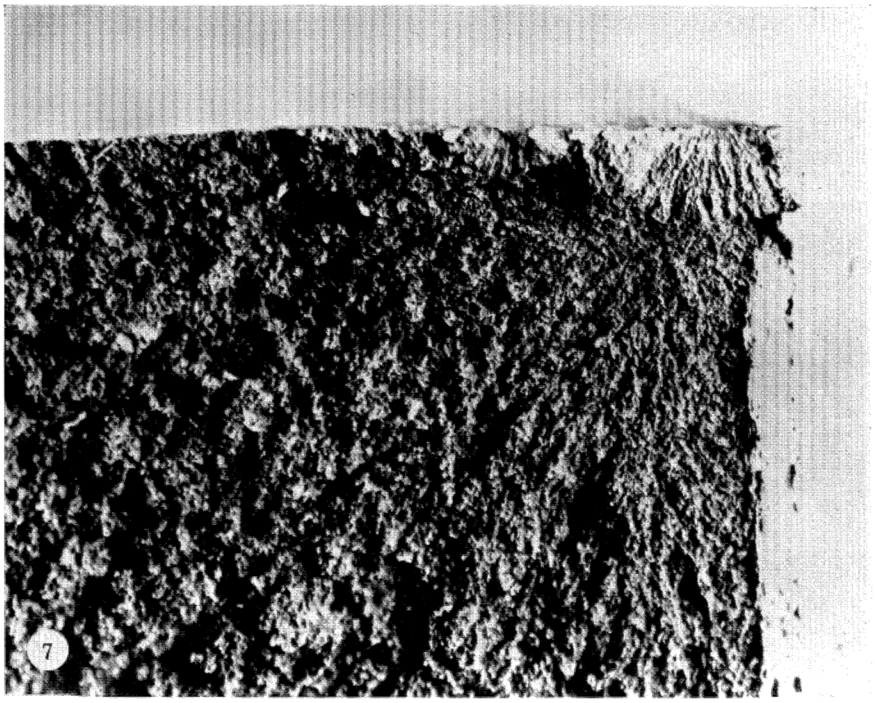

Figure 6. Consequence of stress-corrosion cracking in a structure having a non-redundant load path. Photograph courtesy of H. C. Burnett, National Bureau of Standards.

Figure 7. Stress-corrosion crack (lighter area, upper right hand corner) which triggered the collapse of the Silver Bridge shown in figure 6. The stress-corrosion crack extended downward about 0.075 in (1.9 mm) before changing to brittle fracture. Photograph courtesy of H. C. Burnett, National Bureau of Standards.

has been developed by comparing macroscopic stress-corrosion characteristics with the microstructure (in the most general sense) of the same alloys as elucidated by the many elegant tools available to the modern physical metallurgist. Just as in Dr Rosenhain's day, a division of labour is required, with equally important roles for the physical metallurgist and the macro-mechanical experimentalist.

Electrochemistry

It is easy for the metallurgist or the engineer to assume that control over the physical metallurgy and the mechanics of a stress-corrosion cracking test covers the essentials. But to match professional level metallurgy and mechanics with amateur level corrosion science invites confusion. Note the specimens of the precipitation hardening stainless steel PH 13-8 Mo tested in salt water and shown in figure 5, plate 17. The lowermost specimen was tested at a high K level, which with the participation of the chloride ions ruptured the protective oxide film at the bottom of the precrack and initiated stress-corrosion cracking, which was halted before complete specimen separation. The middle specimen was loaded at a lower K level, and after 1300 h a stress-corrosion crack initiated, but not at the precrack. The initiation was under the wall of the cell which surrounded the central portion of the specimen and contained the salt water. Careful examination of this and of replicate specimens revealed small crevice corrosion pits under the cell wall, which initiated stress-corrosion cracks from an almost smooth surface. Even if these small pits had been as sharp as a fatigue crack, the K level there would have been much lower than at the machined and fatigued notch. The factor which was different in the crevice corrosion pits was that in these stagnant conditions under the cell wall the stainless steel and salt water combined to form hydrochloric acid according to the reaction (Peterson, Lennox & Groover 1970):

$$3\mathrm{Cl}^- + \mathrm{Cr}^{3+} + 3\mathrm{H_2O} = \mathrm{Cr(OH)_3} + 3\mathrm{H}^+ + 3\mathrm{Cl}^-;$$

$$\mathrm{pH} = 1.60 - \tfrac{1}{3}\lg\,[\mathrm{Cr}^{3+}].$$

Thus the low pH in a crevice, due to the hydrolysis of chromium corrosion product, overcame the mechanical disadvantage of the lack of a precrack. The top specimen serves as a test of the effectiveness of electrochemical conditions in crack initiation. Saturated ferric chloride was selected as a convenient way to lower the pH to the range inside an active corrosion pit in the stainless steel, and upon application of the solution to the specimen many cracks initiated immediately from the smooth surface. Hydrochloric acid was found to be equally effective.

The full implications of local solution chemistry to the technological control of stress-corrosion cracking have yet to emerge. But it is abundantly clear that both the understanding and the control of stress-corrosion cracking requires professional level attention from the metallurgist, the applied mechanics specialist, and the corrosion scientist-engineer.

Non-redundant load path designs

There is a class of design which requires a special level of sophistication in knowledge about stress-corrosion cracking and fracture toughness. This class of design is illustrated by the 'Silver Bridge' (figure 6, plate 18), the Hinckley Point A steam turbine (Kalderon 1972), and the rocket motor chamber mentioned above. A study of the collapse of the bridge (Bennett & Harold Mindlin 1973) attributed the failure to the following sequence: A stress-corrosion crack initiated in the eye of an eyebar of which the bridge was built; the stress corrosion may have been due to sulphur compounds in the air. The bridge was built long before the development of fracture mechanics or its application to stress-corrosion cracking, but an after-the-fact test of the steel indicated that in a sulphide solution the critical flaw size to initiate stress-corrosion cracking would be about 0.008 in (0.2 mm), credibly present in the bridge. The fracture toughness indicated that a flaw to set off brittle fracture would need be no larger than 0.05–0.08 in (1.2–2 mm) deep (spread due to inhomogeneous strength). The size of the fatal stress-corrosion crack was actually about 0.075 in (1.9 mm) deep (figure 7, plate 18). Note the generic similarities in the three structures cited: (1) There was stress-corrosion cracking in mild environments, (2) in areas in which the stress was elevated locally (by the eye of the eyebar, by the bore plus keyway of the turbine, and probably by slight out-of-roundness of the chamber), (3) in steels having low crack tolerance, (4) in non-redundant load path designs. The lesson is that if the designer cannot afford a failure, he is denied non-redundant load path designs unless he has complete control over both stress-corrosion cracking and fracture toughness.

CONCLUSIONS

(1) The specifications, standards, and codes needed to control the problem of stress-corrosion cracking require characterization by macroscopic tests. Fracture mechanics offers the possibility of playing a role in this characterization, though probably other methods may also be both valid and needed in special areas.

(2) Understanding the characteristics of a given alloy and the development of improved alloys by other than random methods require the contributions possible through the techniques of physical metallurgy. The macroscopic testing and the physical metallurgical studies are complementary, not competitive, approaches. Both are essential.

(3) The combination of fracture mechanics and other macroscopic tests with physical metallurgy has produced and continues to produce useful descriptive physical metallurgy knowledge about the relation of structure to stress-corrosion cracking characteristics.

(4) For engineering purposes standardized tests are required which can be called out in specifications after the validity of the standardized tests is established.

(5) In the development and validation of standard stress-corrosion cracking tests, the contributions of professional corrosion science are needed to minimize ugly surprises and confusion. In trying to achieve valid standard tests we could do worse than to observe Francis Bacon's maxim, 'Truth will sooner come out of error than from confusion'.

(6) Non-redundant load path designs require exacting attention to fracture toughness and stress-corrosion cracking unless brittle fracture can be either excluded or tolerated.

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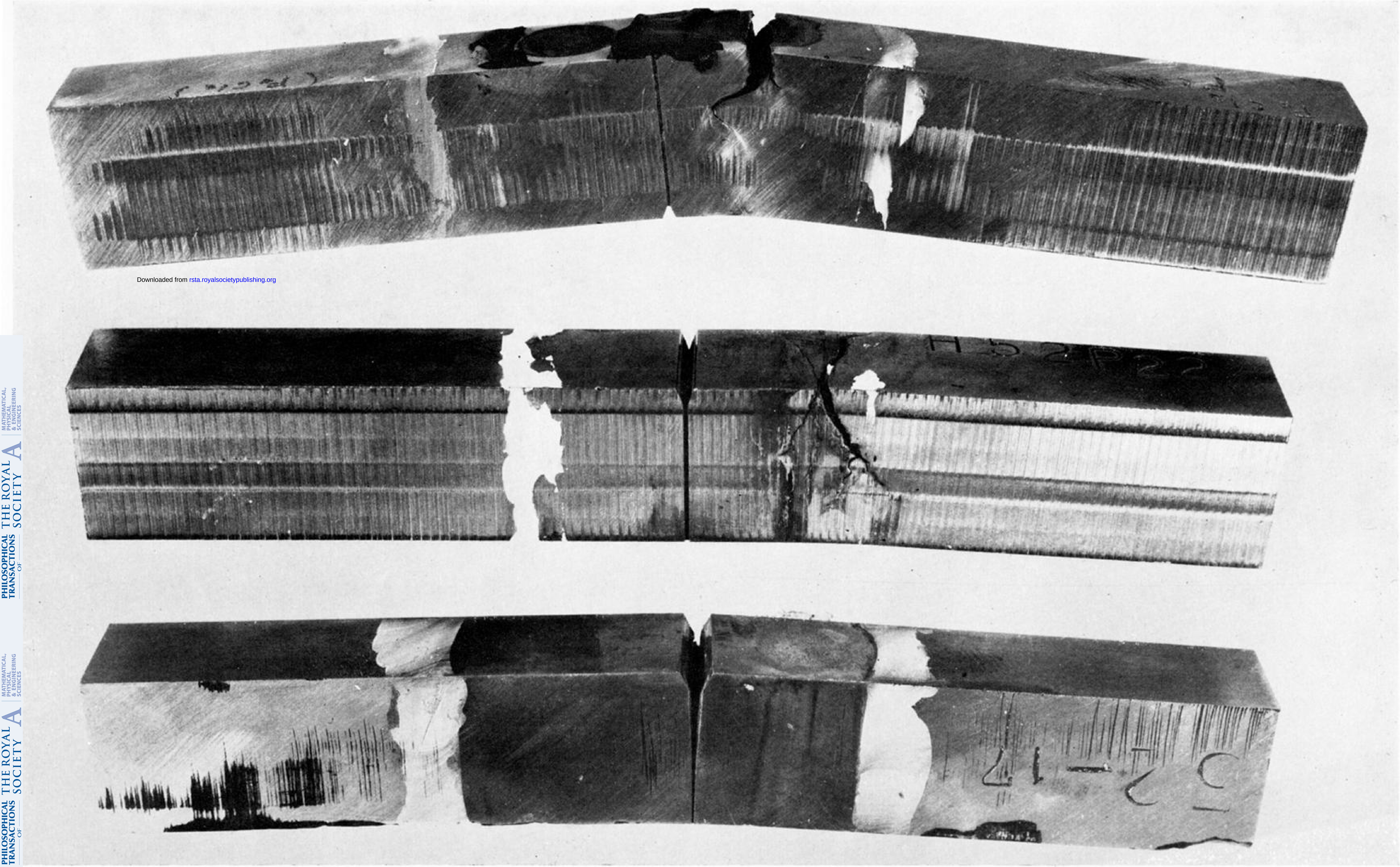

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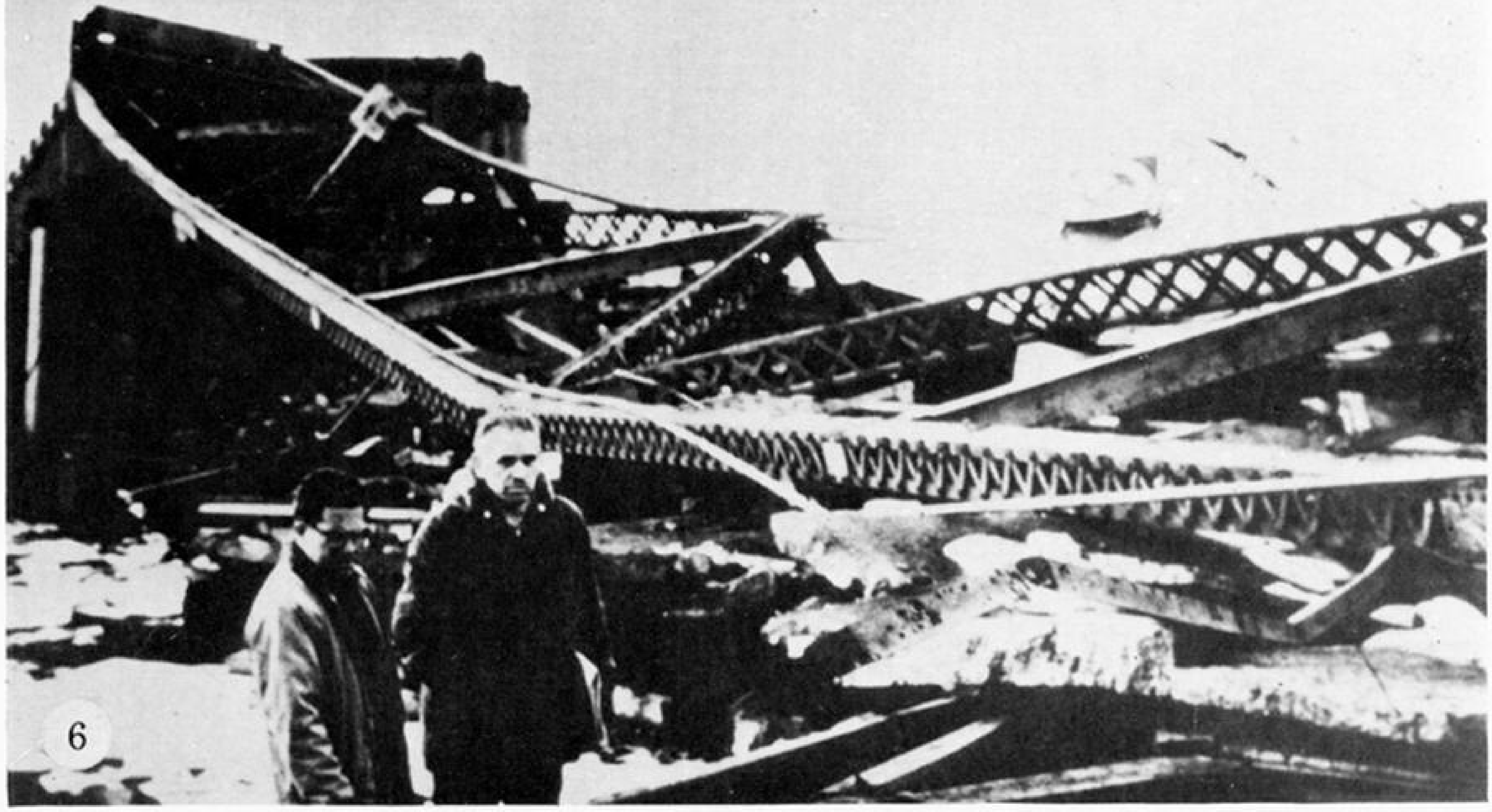

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FIGURE 6. Consequence of stress-corrosion cracking in a structure having a non-redundant load path. Photograph courtesy of H. C. Burnett, National Bureau of Standards.

FIGURE 7. Stress-corrosion crack (lighter area, upper right hand corner) which triggered the collapse of the Silver Bridge shown in figure 6. The stress-corrosion crack extended downward about 0.075 in (1.9 mm) before changing to brittle fracture. Photograph courtesy of H. C. Burnett, National Bureau of Standards.