

# Applying critical potential data to avoid stress corrosion cracking of metals

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Received 8 May 1978

Available data are assembled on the critical potentials for various metal-environment systems below which stress corrosion cracking (SCC) does not occur. Such data can be used practically to avoid SCC by use of cathodic polarization, inhibiting ions or by employing galvanic couples. The slow drift of the corrosion potential before coincidence with the critical potential explains in part the observed induction times for SCC. The relative positions of the two potentials with respect to each other explains the susceptibility or resistance to SCC of various metals in a variety of chemical media. Several models proposed to explain the SCC mechanism are reviewed in the light of the present evidence.

## 1. General aspects

The majority of structural metals, otherwise ductile, are subject to failure by cracking when stressed in tension and exposed to certain environments. This type of failure is called stress corrosion cracking (SCC). It has long been known that cathodic polarization (cathodic protection) can be employed to avoid the initiation of such damage, and even to stop crack propagation, especially if the crack is not too deep.\* What is relatively new information is the observed sharpness of the potential below which such protection occurs, and also the relation of such critical potentials to the mechanism by which various anions act to inhibit SCC.

For example, a stressed 18 wt% Cr-8 wt% Ni stainless steel immersed in concentrated  $\text{MgCl}_2$  solution boiling at  $130^\circ\text{C}$  fails by SCC within less than 1 hour; the time to failure is decreased or increased by applying anodic or cathodic polarization (Fig. 1). The actual critical potential below which time to failure becomes very long, if not infinite, can be determined to within  $\pm 2\text{ mV}$  [1] using a test apparatus previously described [2]. Addition to  $\text{MgCl}_2$  of a few percent of an

extraneous salt like sodium acetate or sodium iodide shifts the critical potential in the noble direction; whenever the potential is shifted noble to the prevailing corrosion potential of the metal in the salt mixture, SCC does not occur and the salt addition acts, therefore, as an effective inhibitor. This mode of behaviour is obviously quite different from that operating for inhibitors of uniform corrosion; it is more nearly in line with the mechanism ascribed to inhibitors of pitting corrosion [3-6].

Similar curves expressing time of failure by SCC as a function of applied potential, or the actual critical potential values, have been reported for a

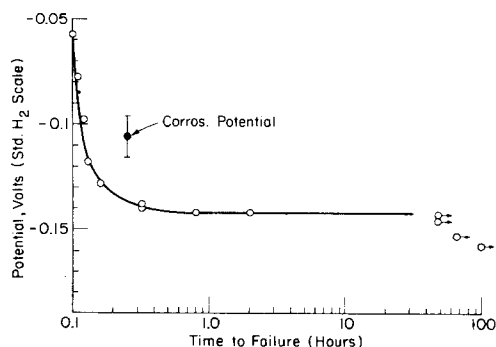


Fig. 1. Effect of applied potential on time to failure of cold-rolled 18 wt% Cr-8 wt% Ni stainless steel in  $\text{MgCl}_2$  boiling at  $130^\circ\text{C}$ .

\* The term stress corrosion cracking as used by introducing interstitial hydrogen into the metal. The latter failures are called hydrogen cracking.

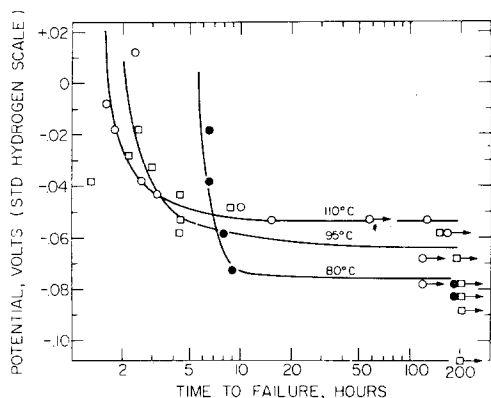


Fig. 2. Effect of applied potential on failure times of 0.09 wt% C mild steel at 3 temperatures in 60 wt% calcium nitrate, 3 wt% ammonium nitrate solution.

variety of austenitic [2] and ferritic [7, 8] stainless steels in  $MgCl_2$ ,  $MgBr_2$  [9] or  $LiCl$  solutions [10], for carbon steels in ammonium carbonate solutions [11] and in nitrate solutions [12–15] (Fig. 2), high strength steels in  $NaCl$  and  $Na_2SO_4$  solutions [16] and in  $NaNO_3$  solutions [17], for an aluminium alloy in 3 wt%  $NaCl$  [18], and for 63 wt% Cu, 37 wt% Zn

brass in copper sulphate–ammonium sulphate solution [20]. A summary of various, but not all, reported critical potentials is listed in Table 1.

For some metals, anodic polarization is sometimes as effective as cathodic polarization, accounting for a narrow potential range in which SCC is pronounced, and outside of which SCC is either retarded or avoided. This situation applies to carbon steels in boiling  $NaOH$  [12, 21, 22] (Fig. 3). A restricted range of potentials in which susceptibility is a maximum is also characteristic of mild steel in carbonate solution [11, 23] and of Ti alloys in  $NaCl$  [19, 24] or  $NaBr$  solution [24].

The importance of information on critical potentials is severalfold. First, it contributes to explaining the induction time for initiation of SCC. Since cracking cannot occur unless the corrosion potential becomes equal to or noble to the critical potential, the time required to achieve coincidence of the two potentials contributes to the delay observed before cracks initiate. This can be demonstrated by immediately polarizing a pickled 18 wt% Cr–8 wt% Ni stainless steel above the critical potential upon immersion in  $LiCl$  sol-

Table 1. Critical potentials for stress corrosion cracking

Alloy	Test solution	Critical potential (V versus SHE)	Reference	
18 wt% Cr–8 wt% Ni, austenitic (WQ)*	$MgCl_2$ boiling at 130° C	–0.128	10	
18 wt% Cr–8 wt% Ni, ferritic (WQ)	$MgCl_2$ boiling at 130° C	–0.098	7	
18 wt% Cr–8 wt% Ni, austenitic (WQ)	$MgBr_2$ boiling at 154° C	–0.04	9	
17 wt% Cr, 2 wt% Ni, ferritic (WQ)	$MgCl_2$ boiling at 130° C	–0.185	7	
25 wt% Cr, 20 wt% Ni, austenitic (WQ)	$MgCl_2$ boiling at 130° C	–0.113	9	
25 wt% Cr, 1 wt% Ni, 3.5 wt% Mo ferritic (ann.)†	$MgCl_2$ boiling at 140° C	–0.16 to –0.19	8	
Al–5.5 wt% Zn, 2.5 wt% Mg	0.5 M $NaCl$ , deaerated, room temperature	–1.11	18	
Ti–8 wt% Al, 1 wt% Mo, 1 wt% V	0.6 M $NaCl$ , room temp.	–0.76	24	
Ti–7 wt% Al, 2 wt% Cb, 1 wt% Ta	3 wt% $NaCl$ , room temp.	–1.1‡	19	
Steel, 0.09 wt% C	60 wt% $Ca(NO_3)_2$ , 3 wt% $NH_4NO_3$ boiling at 110° C	–0.055	12	
0.24 wt% C		–0.073	12	
0.025 wt% N		–0.06	14	
0.013 wt% C		55 wt% $Ca(NO_3)_2$ boiling at 117° C	–0.10	15
0.08 wt% C		4 N $NaNO_3$ , boiling	–0.16	13
0.09 wt% C		170 g $(NH_4)_2CO_3/l$ at 70° C	–0.35	11
0.09 wt% C		35 wt% $NaOH$ , 125° C	–0.85§	12
63 wt% Cu, 37 wt% Zn brass	1 M $(NH_4)_2SO_4$ , 0.05 M $CuSO_4$ , pH 6.5, room temperature	+0.095	20	

\* WQ: water quenched from 1050° C

† Ann: annealed at 980° C, 1 h.

‡ Pre-cracked spec.

§ > 200 hour life. Equal life is observed noble to –0.65 V with less uniform attack.

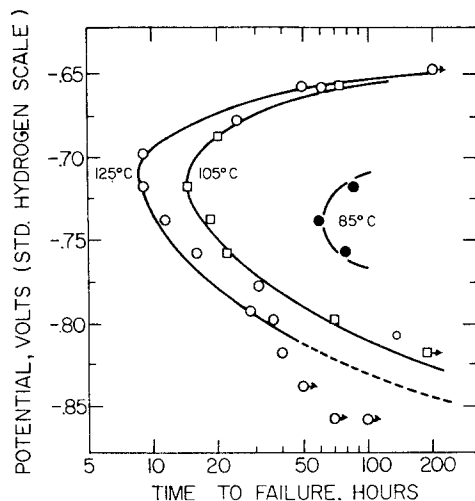


Fig. 3. Effect of applied potential on failure times of 0.09 wt% C mild steel at 3 temperatures in 35 wt% sodium hydroxide solution.

ution, or similarly a 25 wt% Cr–20 wt% Ni stainless steel in  $MgCl_2$  solution, and noting that the induction times for crack initiation are reduced from the order of 1 or more hours to 5 minutes or less [10].

Secondly, recognized inhibiting ions, e.g.  $I^-$  in  $MgCl_2$  solution for 18 wt% Cr–8 wt% Ni stainless steel, or acetate ions in nitrate solutions for carbon steels extend the induction time by shifting the critical potential in the noble direction. Or, more important, if sufficient inhibitor is added, SCC is prevented entirely by a resultant critical potential that lies continuously noble to the normal range of the corrosion potential for any time of exposure.

Thirdly, galvanic couples can be employed to avoid SCC of susceptible metals using an anode metal that is sufficiently negative to displace the potential of the couple below the critical value. In this way, a small area of Ni (corrosion potential =  $-0.18$  V, versus SHE) attached to a stressed 18 wt% Ni–8 wt% Cr stainless steel (critical potential =  $-0.128$  V, versus SHE) in  $MgCl_2$  at  $130^\circ C$  prevents SCC of 18 wt% Ni–8 wt% Cr for any test period studied so far ( $> 200$  hours). Similarly, otherwise susceptible stainless steels do not fail when coupled to differing composition stainless steels having corrosion potentials more active than the critical value [7] (Table 2). The latter observation supports the present proposal that stainless steels consisting of mixed ferrite and austenite

phases of differing composition are resistant to SCC whenever the corrosion potential of the duplex alloy lies below the appropriate critical potentials of the separate phases. For similar reasons, Zn coupled to brass prevents SCC of the latter alloy exposed to ammoniacal solutions, and Zn, Cd or 0.1% Sn–Al coupled to high strength steels prevents SCC of the latter exposed to water or aqueous solutions [16] (Table 2).

It is the shift of critical and corrosion potentials with Ni content that accounts for the susceptibility of cold-rolled ferritic 18 wt% Cr stainless steels containing  $> 1.1$  wt% Ni [7]. Also the observed resistance of ferritic 18 wt% Cr–8 wt% Ni\* in  $MgCl_2$  boiling at  $130^\circ C$  compared to the marked susceptibility of austenitic 18 wt% Cr–8 wt% Ni is a result of the relative order of corrosion potentials and critical potentials and is not related to metallurgical factors as such [7]. In other words, cracks proceed just as readily through either face-centered (austenitic) or body-centered (ferritic) cubic 18 wt% Cr–8 wt% Ni, provided the alloys are polarized above their respective critical potentials. Similarly the observed susceptibility of cold-worked ferritic 18 wt% Cr–8 wt% Ni in  $MgCl_2$  at  $130^\circ C$  but not of the water-quenched alloy is accounted for by differences in critical and corrosion potentials. Some of the relevant data are assembled in Table 3.

For metals which fail within a restricted potential range, environmental conditions must avoid the critical range. For example, stressed mild steel resists failure in boiling concentrated NaOH solutions when pure because the corrosion potential is too negative with respect to the spread of potentials over which failure occurs. Lack of failure in pure NaOH solutions puzzled early investigators of so-called caustic embrittlement of boiler steels. But by addition to NaOH of certain impurities like PbO which shift the corrosion potential into the damaging range, SCC becomes rapid.

\* A laboratory-prepared 18 wt% Cr–8 wt% Ni stainless steel containing  $< 0.01$  wt% C and N transforms spontaneously when quenched from  $1050^\circ C$  to the stable body-centered cubic (ferritic) structure. A laboratory-prepared or commercial 18 wt% Cr–8 wt% Ni containing more C and N quenches to the metastable face-centered cubic (austenitic) structure. When 18 wt% Cr–8 wt% Ni is referred to without designation, the commercial alloy is implied. For properties of ferritic 18 wt% Cr–8 wt% Ni see [7]; also [36–38].

Table 2. Effect of Galvanic Coupling on Stress Corrosion Cracking

Stressed metal	Condition	Solution	Critical potential (V versus SHE)	Coupled to (unstressed)	Corrosion potential of couple (V versus SHE)	Difference (V) (critical — corrosion potentials)	Time to failure (h)	
							Not coupled	Coupled
17% Cr, 2% Ni, bal. Fe	cold rolled	MgCl <sub>2</sub> 130° C	-0.225	16 wt% Cr, balance Fe (equal area)	-0.360	0.135	0.7	> 200
18% Cr, 8% Ni, bal. Fe	cold rolled	MgCl <sub>2</sub> 130° C	-0.145	17 wt% Cr, bal. Fe (commercial) (equal area)	-0.225	0.080	4.9	> 200
4140 steel	quenched, tempered to Rockwell C46	boiling 3 wt% NaCl	-0.416	Zn (double area)	-0.76	0.34	57	> 230
4140 steel	quenched, tempered to Rockwell C46	boiling 3 wt% NaCl	-0.416	Cd (double area)	-0.56	0.14	57	> 230
4140 steel	quenched, tempered to Rockwell C46	boiling 3 wt% NaCl	-0.416	0.1 wt% Sn-Al (commercial) (double area)	-0.55	0.13	57	> 230

Table 3. Effect of critical and corrosion potentials on stress corrosion cracking of stainless steels in  $MgCl_2$  solution

Alloy	Temperature (°C)	Critical potential (V versus SHE)	Corrosion potential* (V versus SHE)	Difference corrosion minus critical potential (V)	Time to failure (h)
18 wt% Cr, 1.1 wt% Ni cold rolled	130	-0.193	-0.198	-0.005	> 200
18 wt% Cr, 2 wt% Ni, cold rolled	130	-0.23	-0.165	+0.065	0.7 ± 0.2
18 wt% Cr, 8 wt% Ni, ferritic, WQ†	130	-0.098	-0.128	-0.030	> 200
18 wt% Cr, 8 wt% Ni ferritic, cold rolled	130	-0.143	-0.138	+0.005	4.9 ± 1
18 wt% Cr, 8 wt% Ni, ferritic, cold rolled	154	-0.060	-0.095	-0.035	> 200
18.8 wt% Cr, 9.2 wt% Ni, austenitic, WQ	130	-0.128	-0.090	+0.038	1.4 ± 0.1

\* steady-state value

† WQ: water quenched from 1050° C

## 2. Interpretation

The overall relevance of critical potentials to the practice of preventing SCC of structural metals employing, for example, inhibitors or galvanic coupling, indicates that any reasonable proposed mechanism must include an explanation for the critical potential. It must also explain why such a potential is usually shifted in the noble direction in the presence of non-damaging ionic species. The major proposed mechanisms for SCC still being discussed are (a) electrochemical (b) surface oxide film fracture (c) hydrogen cracking (d) stress-sorption cracking.

The electrochemical dissolution model of SCC, first proposed in 1940 by Dix [25] attributes the beneficial effect of cathodic protection to a reduced corrosion rate either at the metal surface, thereby preventing crack initiation, or within a crack thereby preventing crack propagation. According to the traditional electrochemical theory of cathodic protection, the corrosion rate of any metal becomes zero at the open-circuit anode potential of the corresponding corrosion cell [26] (Fig. 4). But obviously if the critical potential is identified with the open-circuit anode potential, the corrosion potential of the corresponding corrosion cell must always lie noble to

the critical potential. Since this situation is not realized generally (e.g. iron in nitrate solutions, [12] or high Ni stainless steels in  $MgCl_2$  solution [2]) and it is also possible for many non-oxidizing anions like acetate ions to shift the critical potential noble to the observed corrosion potential, it is not likely that the electrochemical dissolution model can be correct. The critical potential must have another interpretation, as is described later.

The failure of the electrochemical model to explain the critical potential has not been answered by its proponents who criticize alternative models (see, for example, [27]). In addition, it is also apparent that the specifically damaging

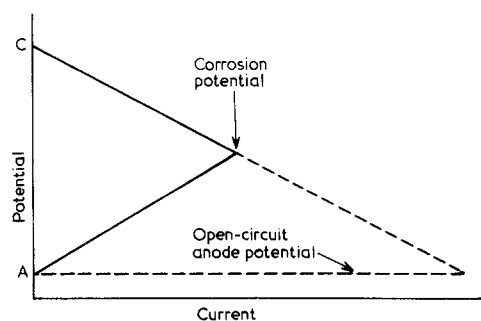


Fig. 4. Schematic polarization diagram for a uniformly corroding metal, illustrating conventional cathodic protection achieved by polarizing to the open-circuit anode potential.

anions causing SCC are unrelated to any level of ionic conductance or of any specific polarization characteristics, both of which enter electrochemical reaction rates. Also there is no reasonable electrochemical explanation for the marked inhibiting behaviour of a small percentage of added salts, e.g. 4 wt% KI added to 36.5 wt%  $\text{MgCl}_2$  solution on the SCC of 18 wt% Cr–8 wt% Ni stainless steel, or the beneficial effect of coupling Ni to 18 wt% Cr–8 wt% Ni, or the observed resistance of ferritic 18 wt% Cr–8 wt% Ni compared to the susceptibility of austenitic 18 wt% Cr–8 wt% Ni stainless steels in  $\text{MgCl}_2$ , or the resistance of 18 wt% Cr stainless steel and the susceptibility of the same alloy containing 2 wt% Ni, or the resistance of cold-worked mild steel but not annealed mild steel to boiling nitrates.

Some investigators have proposed a model based on progressive discontinuous fracture of a passive or oxide film natural to the metal surface, for example as initiated at slip steps, with enhanced corrosion at breaks in the film. A slowly growing oxide film becomes visible when 18 wt% Cr–8 wt% Ni is exposed for long times to  $\text{MgCl}_2$  solution boiling at 130° C, and more so when boiling at 154° C. This thick film, however, which has no obvious relation to the thin passive film normally present on air-exposed stainless steels, accounts mainly for the drift of the corrosion potential in the noble direction [2] and for an extended induction time. But the critical potential to stop crack growth in pre-cracked 18 wt% Cr–8 wt% Ni stainless steel specimens (cracks 0.1–0.25 mm deep) is the same within 5 mV as that to prevent crack initiation in a smooth specimen [10]. Growth of the surface oxide, furthermore, is visible either above or below the critical potential. Since the factors favouring oxide formation and fracture within a crack are expected to differ from those occurring on the metal surface (explaining the SCC of pre-cracked Ti alloys, but not of smooth specimens [28]) and the oxide would crack whether above or below the critical potential, the oxide fracture model does not apply to 18 wt% Cr–8 wt% Ni nor probably to any other type of stainless steel. The above experiment on pre-cracked specimens shows that the critical potential is not sensitive to the stress intensity factor (at least for 18 wt% Cr–8 wt% Ni) or to the stage of crack growth. Accord-

ingly, the rate of repassivation of the alloy at the crack tip, considered to be a critical factor by some investigators who support the electrochemical model, is also apparently not important. Additional weaknesses of the oxide fracture model for any metal–environment system are that it fails to explain the specific nature of damaging ions, e.g.  $\text{NO}_3^-$  (but not  $\text{CrO}_4^{2-}$ ) for steel,  $\text{Cl}^-$  (but not  $\text{I}^-$ ) for stainless steels,  $\text{Cl}^-$  (but not  $\text{NO}_3^-$ ) for Ti alloys, and  $\text{CuSO}_4-(\text{NH}_4)_2\text{SO}_4$  but not  $\text{NH}_4\text{Cl}$  for many copper-base alloys. It would seem that if brittle oxide or passive films played an important role, SCC would be much more widespread among all metals in a variety of environments including pure metals which are typically immune (Fe, Cu, Ti, Al, Zn).

Similarly, a general mechanism dependent on interstitial hydrogen is ruled out by several observations, the most important of which is that cathodic polarization which favours  $\text{H}^+$  ion discharge, and should accelerate damage, actually prevents SCC. Those particular instances where cathodic polarization induces cracking (e.g. stressed high strength steels or martensitic stainless steels) are better described as hydrogen cracking which occurs by a different mechanism. The difference between hydrogen cracking and SCC of ferritic 18 wt% Cr–8 wt% Ni stainless steel can be distinguished by the effect of rolling direction. Hydrogen cracking of the stressed alloy cathodically polarized occurs predominantly when the stress is applied perpendicular to the rolling direction, whereas SCC of this alloy in  $\text{MgCl}_2$  occurs largely independent of rolling direction [29]. The hydrogen model also does not provide an obvious explanation for the sharply defined critical potentials; furthermore, observed values of potential for crack initiation do not correlate with thermodynamic conditions for  $\text{H}^+$  ion discharge. When the pH of  $\text{MgCl}_2$  test solution is lowered by addition of HCl, the observed decrease in failure time of austenitic 18 wt% Cr–8 wt% Ni stainless steel (increased susceptibility) is explained not by more interstitial hydrogen resulting from a higher rate of  $\text{H}^+$  discharge, but by a more noble open circuit cathode potential (the hydrogen electrode potential) which increases the difference between corrosion and critical potentials. A sufficiently oxidizing cation added to the  $\text{MgCl}_2$  solution, e.g.  $\text{Fe}^{3+}$ , acts in the same manner.

For high strength steels, cathodic polarization suppresses cracking in boiling 3 wt% NaCl or 3 wt% Na<sub>2</sub>SO<sub>4</sub> at or below a typical critical potential. Hence galvanic coupling of such steels to more active metals suppresses SCC (Table 2). Cold working of unpolarized high strength steel also suppresses cracking by shifting the critical potential noble to the corrosion potential. Additional cathodic polarization to some potential well below the critical value again causes cracking, but now the failure is one of hydrogen cracking, contrary to the former situation, and cold working the steel now makes it *more* susceptible. Addition of alkali to the NaCl or Na<sub>2</sub>SO<sub>4</sub> solution prevents SCC of the unpolarized steel not because less hydrogen is generated, but rather because of a shift of the critical potential noble to the corrosion potential. The evidence, in other words, supports two differing mechanisms operating in the environmental cracking of high strength steels with differing required preventative measures.

There remains among other possibilities, the mechanism called stress-sorption cracking [30–33]. According to this model, the critical potential is that value above which damaging anions adsorb on appropriate mobile defect sites of the stressed metal accompanied by metal atom decohesion, and below which desorption occurs with the stressed metal remaining intact. In principle, such a potential is well defined, in accord with observed chemisorption phenomena, and it can be specific to the anion and to the metal. The crack initiates where adsorbed damaging anions at a notch or surface irregularity decrease affinity of metal atoms one for the other. In accordance with a physical process, the transgranular crack surfaces (e.g. in 18 wt% Cr–8 wt% Ni) appear as a feathery brittle-type fracture [34]. Anodic as well as cathodic polarization succeeds in preventing SCC of some systems, as described earlier, either because damaging adsorption occurs only within a limited potential range, or because the anodic corrosion products possess inhibiting properties, or rapid metal dissolution blunts a crack nucleus. If passive films are formed they may obstruct adsorption of damaging ions. Which explanation applies probably varies with the metal–environment system.

Inhibiting anions that adsorb but do not reduce metal atom affinities sufficiently to initiate a crack,

displace damaging ions by mass action, thereby making it necessary to increase the applied positive charge of the metal by anodic polarization. Doing so restores the concentration of damaging anions necessary to crack initiation, and shifts the potential in the noble direction. The mechanism is parallel to that operating in pitting corrosion where extraneous anions in general act as inhibitors by moving the critical pitting potential noble to the corrosion potential [6]. There is no evidence, however, that the critical pitting potential which applies to unstressed passive metals is the same as the critical potential for SCC.

Cold working a metal may serve to shift the critical potential in the active direction (e.g. 45 mV in the case of ferritic 18 wt% Cr–8 wt% Ni, cold reduced 35%, Table 3) indicating perhaps greater ease of adsorption on the resulting imperfection sites. Cold working in some instances may entirely erase suitable paths for crack propagation (e.g. by affecting the distribution of interstitial atoms which are necessary to suitable adsorption sites) and hence may eliminate the critical potential (as for carbon steels cold reduced > 50% in nitrate solutions). A small amount of Ni in susceptible ferritic stainless steels (or in iron) is necessary to produce the appropriate defect sites on which damaging adsorption of Cl<sup>-</sup> can occur; in austenitic stainless steels, on the other hand, still larger amounts of alloyed Ni (> 45%) decrease affinity of the imperfection sites for Cl<sup>-</sup> ions. Accordingly, radioactive Cl<sup>-</sup> ions were observed by Bergen [35] to migrate over the surface of a stainless steel toward the areas of maximum stress where favourable defect sites are abundant. The effect was less pronounced for higher Ni content stainless steels, presumably because of the lower affinity of such sites for Cl<sup>-</sup>, in line with the decreased susceptibility of high nickel stainless steels to SCC.

The effect of some interstitial atoms like carbon or nitrogen to increase susceptibility along the grain boundaries of iron, or the presence of nitrogen similarly along the slip planes of austenitic stainless steels, probably prolongs the surface half-life of mobile defects sufficient to allow adsorption, which takes time. In the case of pure metals, the defects presumably move into and out of the metal surface too rapidly for adsorption to succeed, in line with the resistance of pure metals

to SCC. On the other hand, stress corrosion cracking of some metals taking place only above room temperature, as for the stainless steels, probably reflects a balance between increased mobility of imperfections and an increased rate of adsorption.

Many of the above ideas regarding mechanism at the present state of knowledge are obviously speculative. They offer the possibility, however, of explaining many of the complex phenomena of SCC difficult to explain otherwise, and they have the advantage of tying in the basic cracking mechanism of SCC with the parallel stress cracking of specific solid-liquid metal combinations (liquid metal embrittlement) and of various plastics exposed to specific organic solvents, all of which presumably occur by reduction of atomic or molecular bonding. The obvious major gap of information requiring further knowledge is the underlying chemistry of mobile defect sites at which metal decohesion occurs and which can account for  $\text{NO}_3^-$  but not  $\text{Cl}^-$  inducing the cracking of mild steel, and similarly  $\text{Cl}^-$  but not  $\text{I}^-$  inducing the cracking of austenitic stainless steels. This entire area of chemistry, i.e. the chemistry of the deforming metal lattice as contrasted with that of the static metal lattice, is a complete vacuum. If the stress-sorption cracking model is correct, therefore, a whole new area of chemistry is opened up which, in the opinion of the author, deserves concentrated attention and study.

### 3. Conclusions

(a) Observed critical potentials can be usefully applied to the practical avoidance of metal failures by SSC. Failures are not observed at any potential lying immediately active to the critical value. Various extraneous anions added to the environment act as inhibitors because they shift the critical potential noble to the prevailing corrosion potential. Similarly galvanic couples whether applied externally, or resulting from the interaction of multiphase components of an alloy as in some stainless steels, are effective whenever the corrosion potential of the couple lies below the critical potential.

(b) The critical potential-corrosion potential relation explains the susceptibility of ferritic stainless steels containing  $> 1.1$  wt% Ni, also the resistance of ferritic compared to austenitic 18 wt% Cr-

8 wt% Ni stainless steels, and the improved resistance of austenitic stainless steels with increased nickel content.

(c) The electrochemical dissolution model is inconsistent with an observed corrosion potential sometimes active to the critical potential. The model based on fracture of a passive or oxide surface film is not supported by observed critical potentials to stop crack propagation approximating the potentials to stop crack initiation in an 18 wt% Cr-8 wt% Ni stainless steel. The hydrogen mechanism is contradicted by the hydrogen cracking of cold-rolled ferritic 18 wt% Cr-8 wt% Ni but not of austenitic 18 wt% Cr-8 wt% Ni, and the ready SCC of either alloy in  $\text{MgCl}_2$ ; also by the marked effect of rolling direction on hydrogen cracking but not on SCC of ferritic 18 wt% Cr-8 wt% Ni, and by the possibility of hydrogen ion discharge at potentials lying either active or noble to the critical potential.

(d) The critical potential is perhaps best explained by the stress-sorption cracking model. The reduction of metal bond strength (decohesion) by damaging anions adsorbed on appropriate defect sites is similar in mechanism to that which explains liquid metal embrittlement and stress cracking of plastics.

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