

## B.1 “SCC of Sensitized and Non-sensitized Austenitic Stainless Steels and Weldments,” by Peter L. Andresen

This background overview paper provides a foundation for understanding the proactive materials degradation concerns for stress corrosion cracking (SCC) of **wrought, unirradiated austenitic stainless steels in both boiling water reactor (BWR) and pressurized water reactor (PWR) environments, including weldments**. There are separate papers for irradiated assisted SCC of stainless steels, for SCC of cast stainless steels, and for lower temperature, mostly-chloride-related pitting and SCC of stainless steels. There are also papers related to BWR and PWR water chemistry, evolving operational practice, start-up and shutdown, and other considerations that influence SCC.

There is a long history of SCC in stainless steels exposed to high temperature water [1-4], and extensive research work designed to understand the dependencies and underlying causes [1-8]. While the mechanical behavior of stainless steels is not dramatically different at 300 °C than at room temperature, many aspects of corrosion and SCC are quite different. All structural materials (e.g., iron and nickel base alloys) are suitable for service in water environments because they form a protective passive film on the surface. When the film is broken (e.g., by scratching or plastic straining), the “bare surface” corrosion rates are *very high*, decaying in a logarithmic fashion over time as the protective oxide film reforms. Even after long time, the corrosion rate does not decrease to zero, but is sustained at a low, “passive” current density.

The nature of the oxide film on stainless steel is quite different in high temperature water (in this context, defined to be above about 150 °C). Near room temperature, the films on stainless steel are very thin (nanometers) and very protective (in terms of passive corrosion current). However, the films on standard “18Cr-8Ni” stainless steels are not that protective in terms of their resistance to aggressive species like chlorides, and even this limited resistance decreases as temperature increases [9,10]. The breakdown of the film can lead to pitting corrosion, crevice corrosion, stress corrosion cracking, etc., with the primary aggravants being increased chloride (halide), H<sup>+</sup> (lower pH), temperature, and oxidant concentration.

Above ≈150 °C, the oxides become much thicker (hundreds or even thousands of nanometers) and somewhat less protective (both in terms of passive current density and tolerance for breakdown from, e.g., chloride) [11]. The composition of the films is not uniform, either through thickness or from environment to environment [11]. For example, in oxidizing (e.g., traditional BWR) environments, Cr is oxidized from Cr<sup>3+</sup> (e.g., Cr<sub>2</sub>O<sub>3</sub>) to Cr<sup>6+</sup> (e.g., CrO<sub>4</sub><sup>=</sup>), which is soluble in water (Figure B.1.1). Thus, these films have an inner layer, which is Cr-rich and an outer layer low in Cr. In deaerated water, the addition of H<sub>2</sub> reduces the corrosion potential (lowering line (a) in Figure B.1.1, which is the H<sub>2</sub>/H<sub>2</sub>O reaction), which results in an increase in the solubility of Fe<sup>2+</sup> (more in pure water than in PWR primary water, whose pH<sub>T</sub> is ≈ 1.5 units higher). Note that oxidants like O<sub>2</sub> are consumed in cracks and crevices, so those films are somewhat different than exterior films (in oxidizing environments). Both films form oxide crystallites on the surface by re-precipitation, which are typically 0.1 – 10 micrometers in size. The films that form in the presence of oxidants are somewhat more protective (i.e., lower passive current density) [11,12].

While there is not complete concurrence on the mechanism(s) of SCC of stainless steels in hot water, the consensus opinion is that SCC growth occurs by a process involving localized deformation (e.g., at the crack tip), which is a shear process that “breaks” or damages the inner protective oxide and produces an accelerated oxidation process as the film repairs. This was

originally conceived as “slip – film rupture – dissolution,” but there is in fact no requirement for an aqueous electrolyte since steam, air or other gaseous environments produce oxidation, which meets the requirement for the underlying mechanism. Thus, many people refer to the mechanism as “slip – oxidation.” An important element of this mechanism is that it is not fundamentally related to (local) stress per se, but to (local) dynamic strain that damages the protective film. Dynamic strain at constant load (e.g., when cracks grow primarily from weld residual stresses) is sustained because the stress / strain field at the crack tip is redistributed as the crack advances, and this requires local strain. Thus, there is an inherent inter-dependency between dynamic strain that produces crack advance, which in turn sustains the dynamic strain. The slow repassivation process in hot water, long term thermal creep, the interaction among crack advance in adjacent grains, irradiation creep, operating fluctuations (e.g., in temperature or pressure), start-up and shut-down, etc. also help sustain crack advance.

Factors such as simplistic or flawed experiments and extrapolation of room temperature data led to the early presumption that SCC would not occur in non-sensitized, unirradiated stainless steel; or in high purity water; or without oxidants in the water, etc. However, it is now recognized that SCC can occur in essentially any environment-material combination in high temperature water, although the ease of initiation and rate of SCC growth can vary markedly. Thus, the concepts of *thresholds* (e.g., in sensitization, water purity, oxidant level, radiation dose/fluence, etc.) that produce immunity to SCC have given way to an understanding of an inherent susceptibility to SCC that varies markedly with material, environment and stress. However, pragmatically, in many cases there are conditions of use that lead to very long lives.

The traditional view of SCC is reflected in Figure B.1.2, which expresses the need for a confluence of stress, environment and metallurgical parameters to exist for SCC to occur. (Drawn today, the central over-lapping region of SCC susceptibility would be much larger). The various parameters that control the initiation or growth rate do not operate independently but rather inter-dependently; thus, some factors (e.g., 10 – 30 ppb sulfate) can have a huge influence under some conditions (e.g., in oxidizing water), but not under other conditions (e.g., deaerated / hydrogenated water). Many such inter-dependencies exist in SCC; indeed, to some extent the effect of essentially all factors are influenced by all other factors. Thus, focusing on the engineering factors that influence SCC can produce a confusing variety of observations, and successful understanding and prediction of SCC must be built on an accurate image of the “crack tip system.”

The primary factors that control SCC of stainless steels in hot water are:

- Degree of sensitization. Sensitization results from Cr carbides that form during thermal exposure (from heat treatment or welding) in the range of 550 – 750 °C, although Cr carbide nucleation can occur below 400 °C in cold worked stainless steels, and growth of existing Cr carbides can occur below 300 °C. The fundamental phenomenon relates to the formation of Cr carbides (usually Cr<sub>23</sub>C<sub>6</sub> in stainless steels), which nucleate preferentially in the grain boundaries. Because C diffuses much faster than Cr, a Cr depletion profile is created adjacent to the grain boundary (the diffusivity of Cr within the grain boundary is much higher than within the grain, so the Cr concentration is reasonably constant along the grain boundary). At higher temperature, the Cr profile is deeper and wider. When the carbon is consumed, the carbide stops growing, and the Cr depletion profile eventually vanishes. This is difficult to accomplish in stainless steels, but such “healed” microstructures can be fairly readily produced in nickel-base alloys.

The effect of Cr depletion is most evident in oxidizing and/or pH-shifted water chemistries (oxidizing environments create a pH shift in cracks and crevices). In oxidizing (e.g., traditional BWR) environments with impurities (e.g., 50 – 100 ppb sulfate), the presence of sensitization can produce a > 20X increase in crack growth rates, and a similar acceleration in crack initiation. However, in deaerated, near-neutral pH water, Cr depletion plays a much lesser role; indeed, the presence of carbides in the grain boundary reduces SCC susceptibility, apparently by making deformation in the grain boundaries more difficult.

Nucleation of Cr carbides is greatly delayed in time, e.g., by reducing the C content (L-grade stainless steels) and adding Mo (type 316 stainless steels). Thus, modern stainless steel components are either not welded or are fabricated from L-grade stainless steels.

Cr depletion also develops during irradiation due to *radiation-induced segregation*. The Cr profiles are much narrower, and the minimum Cr level is generally higher – usually in the range of 12 – 14% Cr for stainless steels containing  $\approx$  18% Cr. The report on irradiation assisted SCC goes into this phenomenon in more detail.

- Oxidants and Corrosion Potential. The presence of oxidants like dissolved  $O_2$ ,  $H_2O_2$ , and Cu ion can increase SCC and corrosion fatigue (CF) growth rates markedly. Oxidants react on metal surfaces and elevate the corrosion potential. As oxidants diffuse into cracks and crevices, they are consumed (electrochemically balanced by reaction with  $H_2$  or metal corrosion). Thus, the interior of cracks and crevices are at low corrosion potential, and the difference in corrosion potential drives migration of anions (like chloride) into the cracks, and cations (like  $H^+$  and  $Na^+$ ) out of the cracks. The effect on SCC and CF of a complex mix of oxidants (and reductants) is fully captured by their effect on corrosion potential. The corrosion potential is not linear with oxidant concentration, and small (ppb) levels of oxidants can produce large (> 300 mV) change in corrosion potential.

Figure B.1.3 shows an example of the effect corrosion potential of SCC. While its effect is very strong on sensitized stainless steels, it also affects non-sensitized stainless steel (and nickel-base alloys, and carbon and low alloy steels). However, SCC growth occurs at moderate rates in deaerated water (and without any prior exposure to water containing oxidants) and so is also a potential concern in PWRs if cracks can initiate.

- Water Purity and pH. Water purity has a profound effect on SCC and CF, and in oxidizing water there is sensitivity to  $\approx$  10 ppb levels of impurities. The most damaging impurities are chloride and sulfate, but most impurities are damaging. In oxidizing environments, even buffering chemistries (e.g., the B/Li chemistries used in PWRs) produce accelerated growth rates (compared to pure water) in oxidizing environments. Because the oxidizing conditions produce a pH-shifted chemistry in cracks and crevices, there is somewhat less sensitivity to bulk pH (for crack growth – crack initiation is more directly dependent on the bulk chemistry).

In deaerated water (where essentially no difference in corrosion potential occurs in cracks), the sensitivity to impurities is low, although levels > 1 ppm can accelerate SCC. The relatively minor shifts in pH associated with pure (deaerated) water ( $pH_{300C} \approx 5.63$ )

and various standard levels of B (as  $\text{H}_3\text{BO}_3$ ) and Li (as  $\text{LiOH}$ ) ( $\text{pH}_{300\text{C}} \approx 6.8 - 7.4$ ) have a relatively minor effect on SCC growth rates.

The addition of  $\text{H}_2$  to deaerated water produces a small decrease in corrosion potential (59.3 mV per 10X change in  $\text{H}_2$  at 325 °C). This is very important for nickel-base alloys, because it can produce a transition across the Ni/NiO phase stability. However, there is pragmatically no way to cross the Fe/ $\text{Fe}_3\text{O}_4$  phase boundary by adding  $\text{H}_2$ . There is evidence that increasing  $\text{H}_2$  produces somewhat enhanced SCC susceptibility, but this has been observed in slow strain rate tests on smooth specimens, and may be associated with accelerated initiation under these conditions – a broadly parallel observation to enhanced corrosion fatigue initiation in stainless steels at low potential vs. high potential.

- Yield Strength / Cold Work / Weld Shrinkage Strain. There are multiple factors that can increase the yield strength in materials, including cold work, irradiation, precipitation hardening, etc. It appears that all have a similar effect on SCC growth rate. For stainless steels, the primary factors are cold work and irradiation, and irradiation is covered in a separate report on irradiation assisted SCC. Cold work occurs as bulk cold work; surface cold work from machining, rolling, grinding, etc.; and weld shrinkage strain in the heat-affected zone adjacent to welds. The latter factor has only recently been recognized and quantified, and equivalent room temperature strains of >20% are often observed near the fusion line. This explains the shift in location of most cracks from  $\approx 6 - 8$  mm (depending on wall thickness) from the fusion line in sensitized piping compared to  $\approx 1 - 3$  mm for unsensitized piping observed in BWRs.

As materials are cold (or warm) worked, the effect of yield strength increases, and its effect on SCC growth rates appear to be captured by yield strength. The presence or absence of martensite in the deformed structure is small, at least for crack growth in high temperature water; martensite might affect crack initiation and lower temperature SCC response (e.g., < 150 °C). Figure B.1.3c shows the effect of cold work. Figure B.1.4 shows that the increase in growth rate is not linear with yield strength, and that yield strength affects SCC in a similar fashion at low and high corrosion potential.

- Temperature. In the range of 250 – 350 °C, increasing temperature increases SCC growth rates. Equally importantly, it appears to help sustain SCC growth. At lower temperature – all things being equal – the growth rates may continue to decrease; but “all things” are never equal. In particular, the corrosion potential changes with temperature; this is especially pronounced and important in the presence of oxidants (Figure B.1.5). Pragmatically, as components cool down, many are exposed to unusual loading and/or water chemistry. BWR components vary only between 274 and 288 °C, where the difference in growth rates is limited. In PWRs, the temperature ranges from 286 °C (core inlet) to 323 °C (core outlet) to 343 °C (pressurizer), and the crack growth rate varies more significantly.
- Stress and Stress Intensity Factor. Stress and stress intensity factor obviously play a large role in SCC initiation and growth. In general, few components are designed for use above a nominal stress  $\approx 80\%$  of the yield strength. The effect of stress intensity factor on crack growth rate appears to vary with water chemistry, and tends to be in the range of  $K^2$  to  $K^3$  (Figure B.1.6). Many cracks grow adjacent to welds, where the weld residual stress profile must be accounted for along with operating stresses. The weld

residual stress profile combined with the inherent effect of crack length,  $a$ , on stress intensity factor ( $K \propto \sqrt{a}$ ), causes a large variation in  $K$  vs. crack length – sometimes resulting in an increase in  $K$ , and sometimes in a  $K$  that rises, then drops to zero.

- Low frequency vibratory loading, high frequency “ripple” loading, and even occasional load perturbations can significantly accelerate SCC growth rates. The distinction between SCC and corrosion fatigue (CF) is poorly demarcated, and with good reason: in most if not all cases, SCC and CF represent a continuum in the *environmentally assisted cracking* spectrum, with increasing strain rate causing higher propagation rates but a lower factor of improvement for SCC mitigation approaches such as water chemistry or material modifications (Figure B.1.7).

$K_{ISCC}$  was once considered to be an invariant material property. However, there is extensive evidence that “ $K_{ISCC}$ ” depends on many parameters – indeed, the evidence of “ $K_{ISCC}$ ” has been based on non-optimal experiments (e.g., use of transgranular fatigue pre-cracked specimens as a basis for growth or non-growth of intergranular stress corrosion cracks). Other tests were performed under decreasing stress intensity factor conditions where the change in  $K$  was large and/or the increment of growth small (e.g., compared to the size of the plastic zone). More recent data has shown well-behaved crack growth rates as low as  $5.5 \text{ MPa}\sqrt{\text{m}}$ . It is impossible to prove that a  $K_{ISCC}$  *doesn't exist*, because one can always choose an incrementally smaller  $K$ , and the growth rates become too small to study in a realistic time frame, e.g., below  $3 \text{ MPa}\sqrt{\text{m}}$ .

### **Predictability of SCC of Stainless Steels**

A strong qualitative understanding and good quantitative predictive capability exists for SCC of stainless steels, esp. for BWR water chemistries and temperatures where extensive SCC has occurred, initially in sensitized pipe weld heat affected zones, and later in unsensitized vessel internal components (while there is a well-behaved continuum between unirradiated and irradiated stainless steels, SCC has occurred in stainless steel components that receive essentially no radiation damage (e.g.,  $<0.01 \text{ dpa}$ , such as the core spray lines).

The distinction between BWR and PWR primary operating conditions is not nearly as great as once thought, esp. as BWRs shift toward low corrosion potential operating by adding  $\text{H}_2$  and, far more effectively, introducing NobleChem™ (which creates a sub-monolayer of Pt or Pt/Rh on all wetted surfaces [13-15]. Under such conditions, the primary differences between BWR and PWR primary water chemistry are coolant additives (typically  $\text{H}_3\text{BO}_3$  and  $\text{LiOH}$ ) that shift the pH at temperature from 5.6 to  $\approx 7.2$ ;  $\text{H}_2$  fugacity ( $\approx 50$  vs.  $3000 \text{ ppb H}_2$ ); and temperature ( $274 - 288 \text{ }^\circ\text{C}$  vs.  $286 - 343 \text{ }^\circ\text{C}$ ). Of these, temperature may be the most important factor for stainless steels; for nickel alloys (where the Ni/NiO transition can be traversed) both temperature and  $\text{H}_2$  fugacity are important [4-8].

Examples of crack growth predictive capability are shown in Figures B.1.3, 4, 6 and 7 for sensitized and cold worked stainless steel [6,16-19]. Cold work is a particular concern – not only are some components used in a cold worked state (such as PWR baffle bolts), but most fabricated components have a surface layer of deformed, hardened material. Shrinkage strain during welding also produces a residual strain profile in the heat-affected zone (and the weld metal) in addition to the more-thoroughly studied residual stress profile [6,16-19]. These strains usually peak at the weld fusion line, generally at an equivalent room temperature cold work level of 15 – 20% (but sometimes higher) (Figure B.1.8).

Quantifying and understanding SCC in most systems relies predominantly on high quality, reproducible, consistent SCC data [20,21]. This has proven difficult enough in crack growth rate measurements, but is generally more difficult in crack initiation experiments as well as evaluation of plant data.

It is important to understand that the effect of individual changes (such as corrosion potential, water purity, temperature, cold work, stress intensity factor, irradiation, etc.) on SCC cannot be viewed in isolation in most experiments, and rarely if ever in plant components. For example, the effect of BWR water purity is dramatically different at moderate to high corrosion potential than at low corrosion potential (as indicated by the predicted curves in Figure B.1.3). Similarly, the factor of improvement observed for various mitigation techniques varies with loading and water chemistry conditions (Figure B.1.7).

It must also be recognized that there is a time-based evolution (e.g., related to plant operating conditions, or to radiation damage – as addressed in the IASCC topical report) and a crack-depth based evolution (e.g., in residual stress, stress intensity, cold work, sensitization, microstructure...), and these produce complex changes in predicted and observed response. Figure B.1.9 shows an example of this interaction in terms of the predicted difference in crack growth trajectory vs. time in different welds. Figure B.1.10 shows an example of crack length vs. time observations and predictions for sensitized type 304 stainless steel pipes.

### ***SCC Mitigation***

There are many approaches that have been pursued and adopted to mitigation SCC in stainless steels. In BWRs, eliminating sensitization (grain boundary Cr depletion) was a primary focus in the 1970s, and most BWRs replaced all recirculation piping with grades of stainless steel that resisted weld sensitization (by lowering C, adding Mo, using lower heat input, etc.). Efforts were also made to reduce weld residual stress by last-pass heat sink welding, induction heating, and mechanical stress improvements. Some BWRs operated with high aqueous impurity levels (esp. chloride and sulfate), and major efforts were undertaken to improve water purity.

The most effective mitigation strategy for existing plant components is to reduce the corrosion potential (Figure B.1.3), and the most effective way to accomplish this is using electrocatalysis. The techniques to make surfaces electrocatalytic are numerous, but the most effective and economical approach is NobleChem™ [13-15], which involves the injection of ionic forms of Pt (or Pt and Rh), which electrolessly reduce and deposit on the surfaces of all wetted parts in BWRs. Applications have been performed on about 30 BWRs, and a new on-line application technique will be performed at a lead plant in 2005.

There is a possibility of attaining further SCC mitigation by adding Zn [22,23]. This is applicable to both BWRs and PWRs, and because Zn is a cation, it is most effective if the corrosion potential is low (i.e., BWRs operating with NobleChem™). Accurate quantification of the benefit of various Zn levels must still be performed and validated. Other approaches are more achievable only in new plants or if components are replaced. For example, the presence of grain boundary carbides or other particles impedes crack advance provided that they are not accompanied by Cr depletion [16,17].

### ***SCC of Stainless Steels – Concerns and Emerging Issues***

A number of ambiguities and emerging concerns exist in the area of SCC of stainless steels. Despite some improvements, the quality of experimental crack growth rate data is still a large factor in the observed scatter and disagreements over the quantitative effect of specific parameters. Even more problematical is the lack of statistical confidence and even qualitative insight into the controlling factors for crack initiation, although in general the same factors that enhance the crack growth rate also accelerate crack initiation.

Among the emerging concerns is that role of increasing stress intensity factor ( $K$ ) as the crack grows ( $dK/da$ ) [24]. There is usually a large positive  $dK/da$  early in the crack growth process because  $K \propto \sigma\sqrt{a}$  (stress times the square root of crack depth), and the integrated effect of the weld residual stress profile produces an increasing stress at the crack as the crack grows. There continues to be a change in  $K$  as the crack grows longer, but the magnitude of the  $+dK/da$  or  $-dK/da$  is smaller (Figure B.1.11). Unfortunately, the few studies that have evaluated dynamic changes in  $K$  have been performed using a fixed change in load ( $dP/dt$ ) or displacement vs. time (similar to  $dK/dt$ ). However, this is expected and observed to yield non-conservative response because it does not produce the accelerating effect of *positive feedback* as the crack begins to grow faster, causing  $K$  to increase faster, causing the crack to grow faster... (Figure B.1.11) Conversely, with decreasing  $dK/da$ , as the crack slows, the rate of change of  $K$  slows, causing further slowing in the crack growth rate... At plant-relevant values of  $-dK/da$ , stable, well-behaved SCC can be sustained from 30 MPa $\sqrt{m}$  to below 12 MPa $\sqrt{m}$  [16,17,25]. Using  $-dK/dt$  (or  $-dP/dt$ ) fails to provide the important feedback between the rate of change of  $K$  and the rate of crack growth, and tends to produce crack arrest.

Another concern is the role of Si, which has been shown to cause elevated crack growth rates and a limited effect of stress intensity factor or corrosion potential (Figure B.1.12) [16,17,25]. While this may be a particular concern for irradiated materials, many stainless steels have a nominal Si content of 0.7 – 1.0%, which may be sufficient to cause elevated growth rates. Si readily oxidizes and is quite soluble in high temperature water – indeed, it is typically present in BWR (and probably PWR) water at levels about 100X higher than other impurities (typically 100 – 1000 ppb). It does not affect conductivity because it dissolves primarily in non-ionic form. A more speculative concern is the role of Mo, esp. in type 316 stainless steels, which have 2 – 3% Mo. Mo does not impart large improvements in corrosion or stress corrosion resistant in high temperature water (as it does below  $\approx 100$  °C), and it does readily oxidize and become soluble. Importantly, many Type 316 stainless steels show very high Mo segregation at the grain boundary (10 – 20% Mo), the extent of which peaks at certain cooling rates following annealing [26]. This could have a significant effect on SCC response, although no studies have yet been performed to confirm it.

A final concern relates to the role of environment in fracture toughness data. Essentially all of the fracture toughness data obtained above 200 °C were obtained in air. It is well established that high crack propagation rates under constant load – and reduced toughness in J-R tests – are observed in the range of about 75 – 140 °C for precipitation hardened Ni alloys [27-29], and it is reasonable to suspect that this might occur in stainless steels and Ni alloys like alloys 600 and 690 (esp. if their yield strength is elevated from cold worked, weld shrinkage strain or irradiation). Very preliminary data at 288 °C on cold worked stainless steel showed that the specimen unexpectedly failed as the  $K$  was allowed to increase to about 88 MPa $\sqrt{m}$  at the end of a test. The load was accurately known, as was the crack depth from post-test fractography at the point sudden failure occurred [17].

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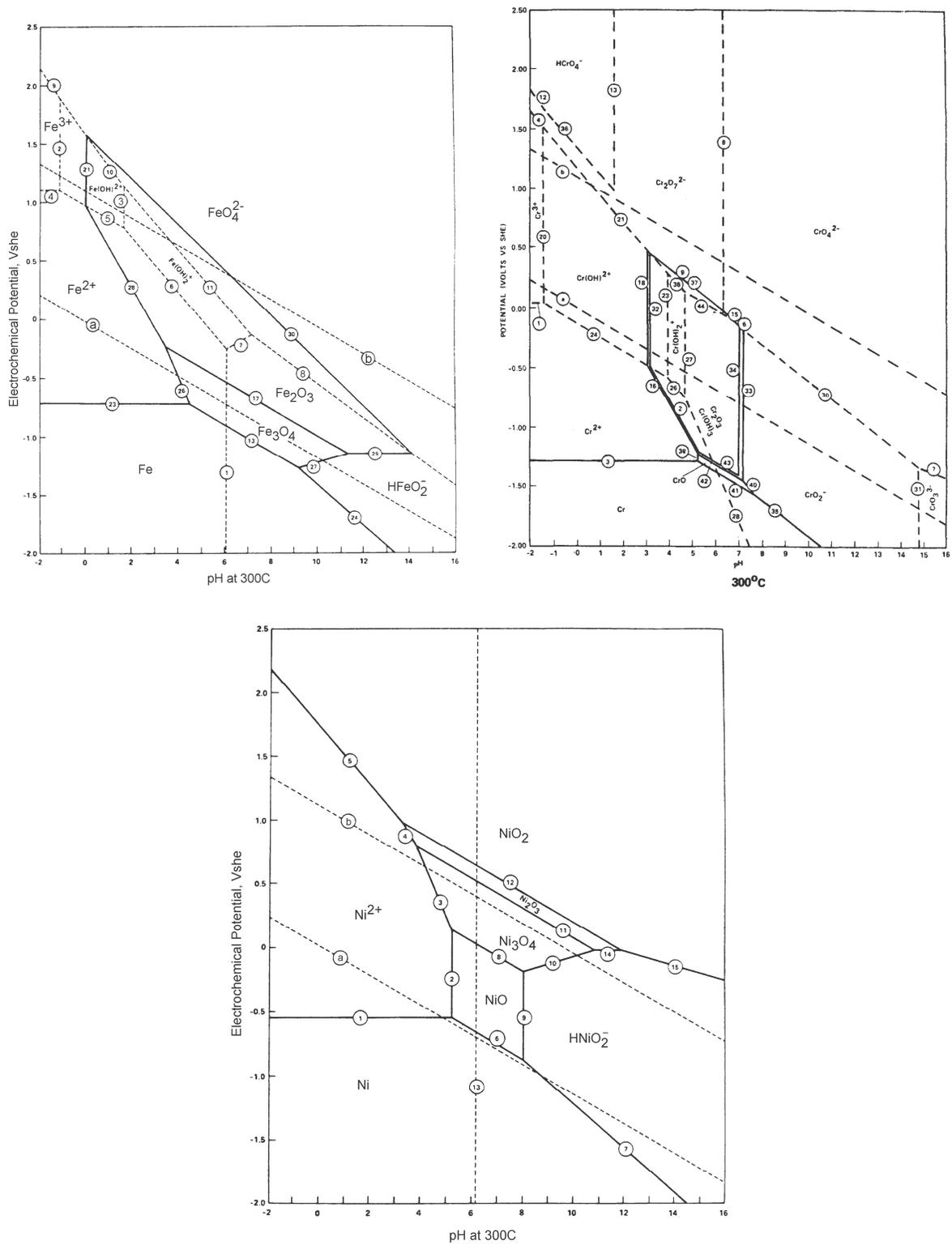


Figure B.1.1 Pourbaix diagrams for Cr, Fe and Ni in 300 °C water [30]. Used by Permission of EPRI.

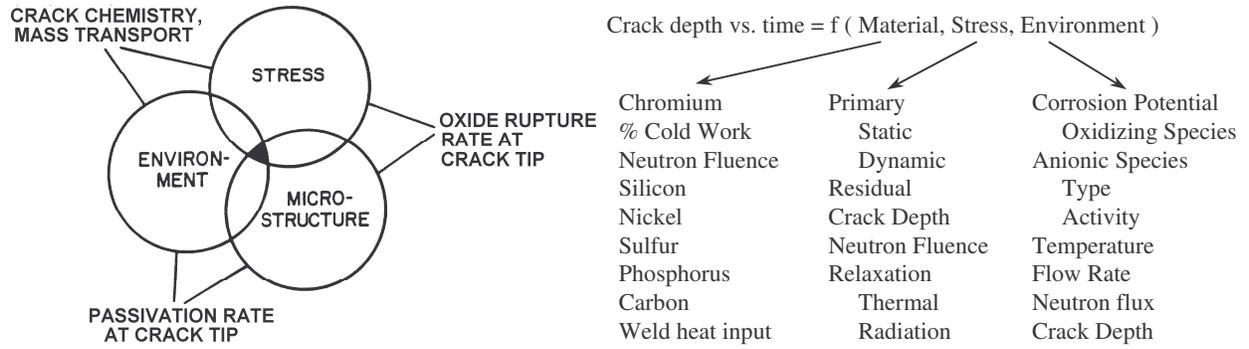
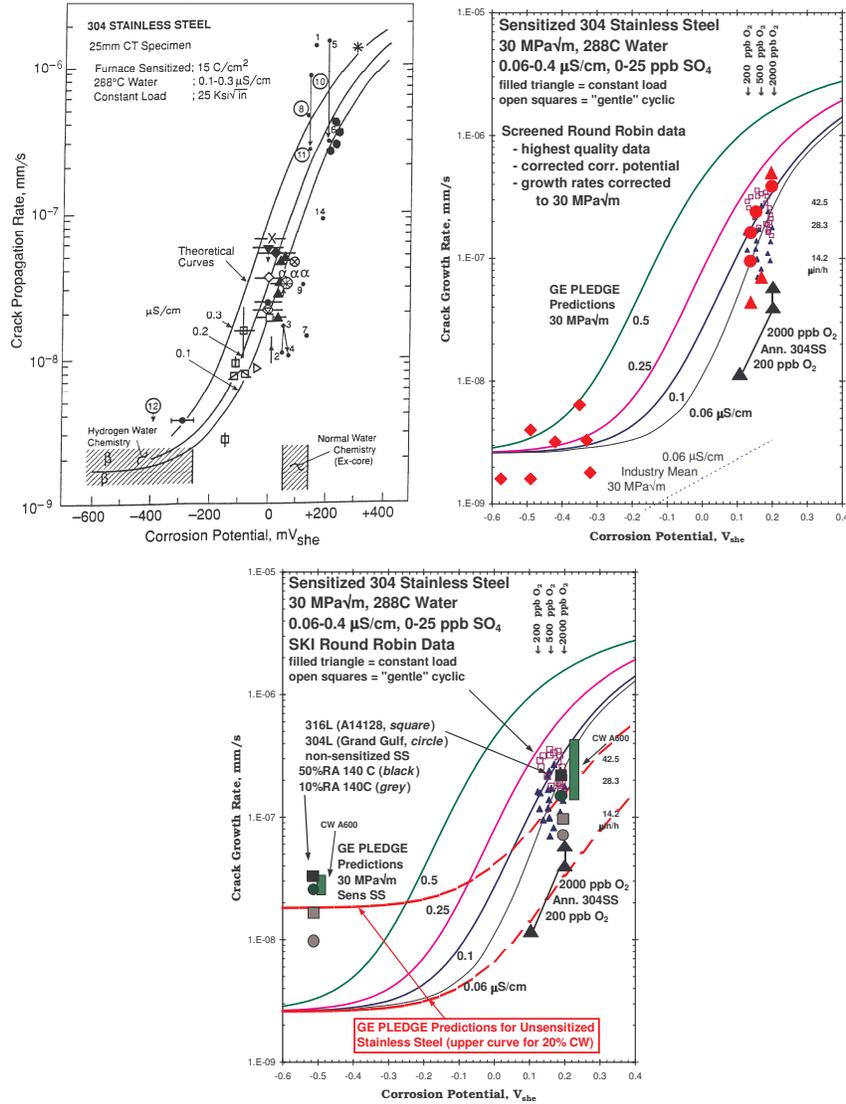
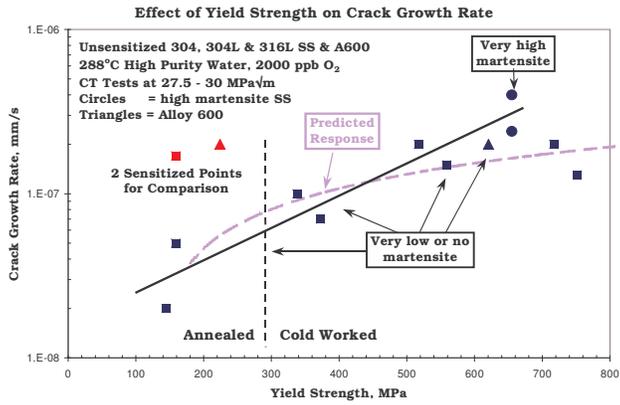
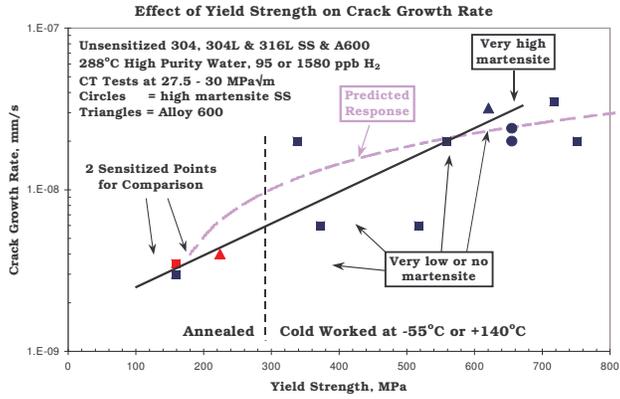


Figure B.1.2 The complexity of SCC is reflected in the large number of influential variables and the associated requirement that all 20 to 40 in a given system be adequately controlled [4,5].  
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(a) (b) (c)

Figure B.1.3 SCC growth rate vs. corrosion potential for stainless steels tested in 288 °C high purity water containing 2000 ppb O<sub>2</sub> and 95 – 3000 ppb H<sub>2</sub> [4,5,16,17]. (© 2003 by The American Nuclear Society, La Grange Park, Illinois)



(a)

(b)

Figure B.1.4 Effect of yield strength and martensite on the stress corrosion crack growth rate on stainless steel and alloy 600 in 288 °C, high purity water (<0.10  $\mu$ S/cm outlet) at (a) low and (b) high potential [16,17] © 2003 by The American Nuclear Society, La Grange Park, Illinois

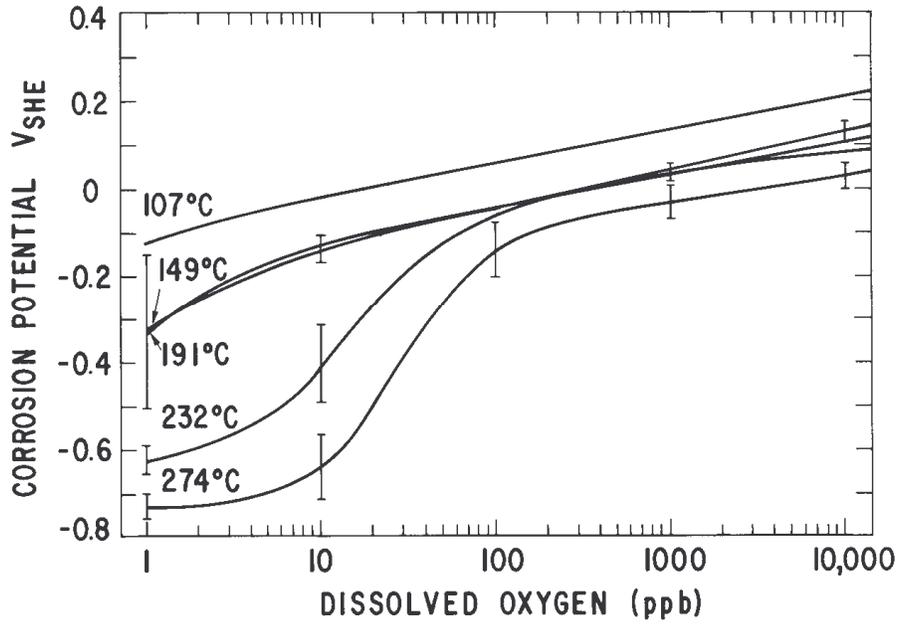


Figure B.1.5 Corrosion potential vs. dissolved O<sub>2</sub> and temperatures for stainless steels in pure water [31].

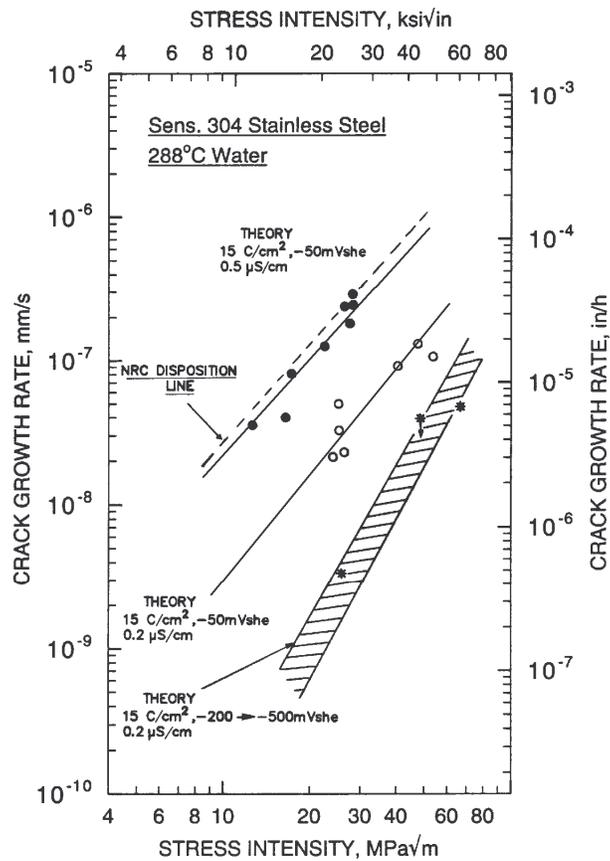


Figure B.1.6 Effect of stress intensity factor on the SCC growth rate of sensitized type 304 stainless steel in high temperature water [4,5]. (Reprinted with Permission from Elsevier)

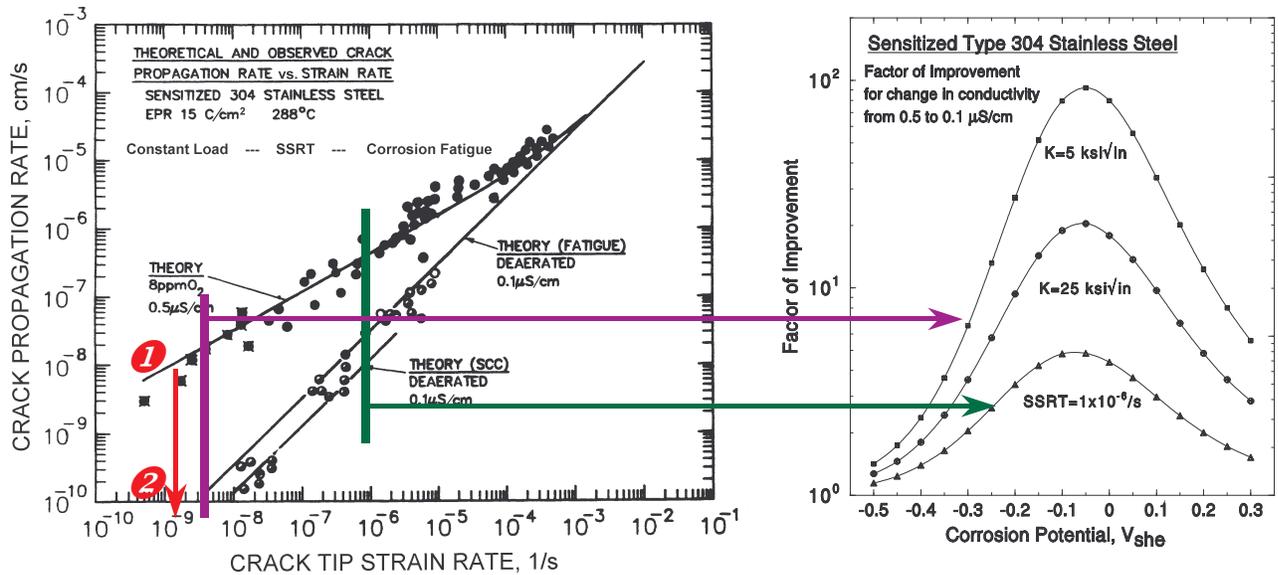


Figure B.1.7 Crack growth rate vs. crack tip strain rate showing the ability to predict environmental cracking across a range of constant load/K, slow strain rate and corrosion fatigue response. When water chemistry or the material is changed, the resultant curves are not parallel, but diverge at lower crack tip strain rate. Thus, the benefit observed in a laboratory test or component will depend on the testing condition, both loading and water chemistry [4-6]. (Courtesy of Taylor and Francis Group)

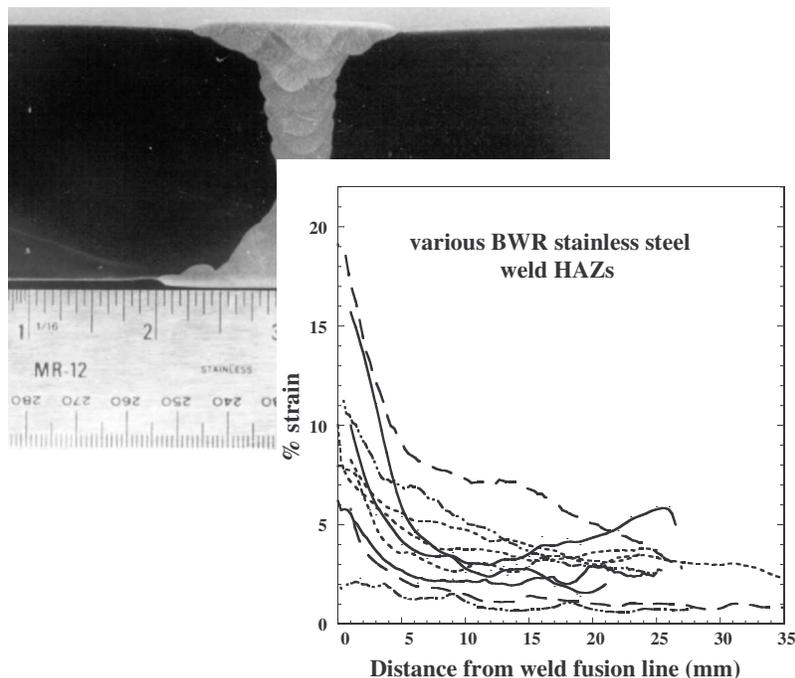


Figure B.1.8 Weld residual strain vs. distance from the weld fusion line for stainless steel welds. If the number of welding passes is limited, the peak residual strain can be below 10% equivalent room temperature tensile strain. However, most pipe welds that have been analyzed show residual strains in the range of 15 – 20%, with some slightly above 25%. The residual strain is also highest near the root of the weld [18]. (Reprinted with Permission from TMS)

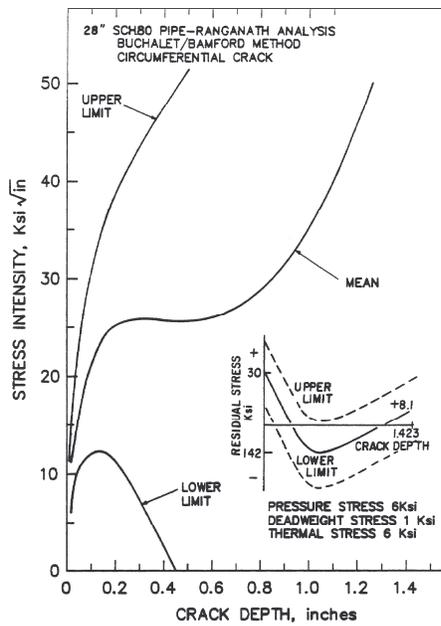
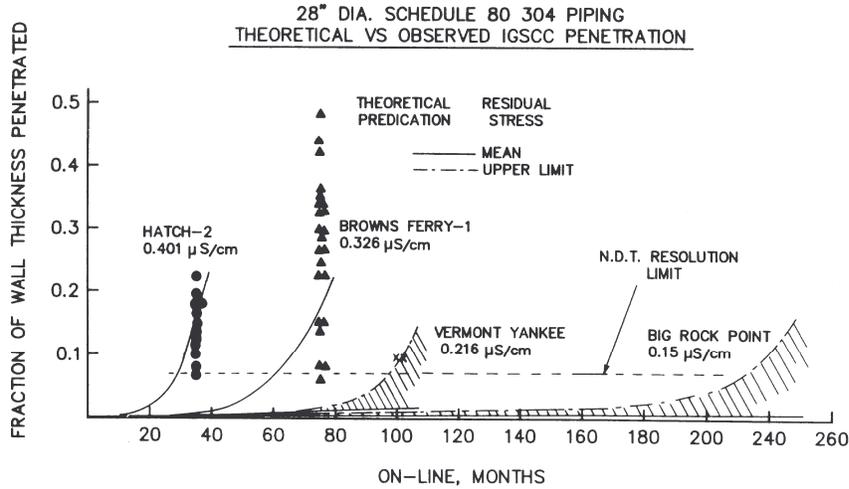


Figure B.1.9 (a) Crack length vs. time for pipe welds in various BWRs whose water purity varied markedly during their first years of operation. (b) Stress intensity factor vs. crack depth for pipe welds. The complex changes in residual stress, stress intensity factor and crack length vs. time that results in plant components. Irradiation makes the situation much more complex (see IASCC topical paper).

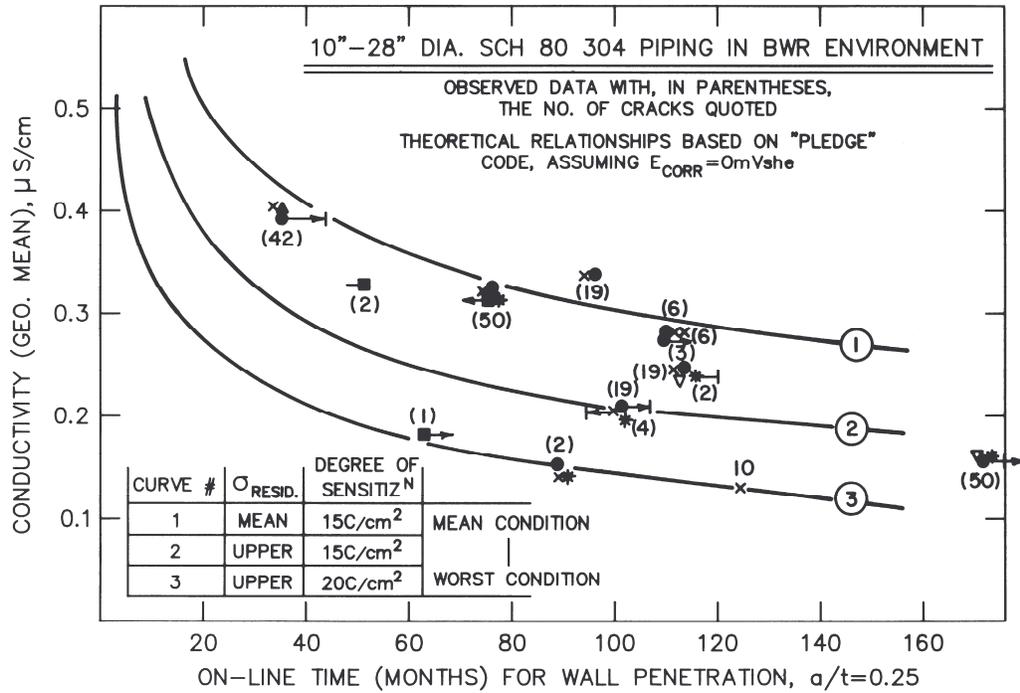


Figure B.1.10 The effect of conductivity on on-line months to achieve a crack depth of 25% of through-wall [4,5]. (Reprinted with Permission from Elsevier)

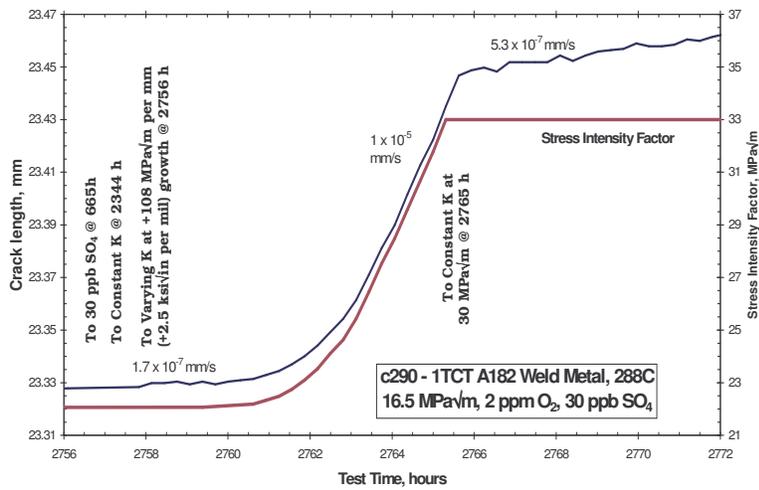
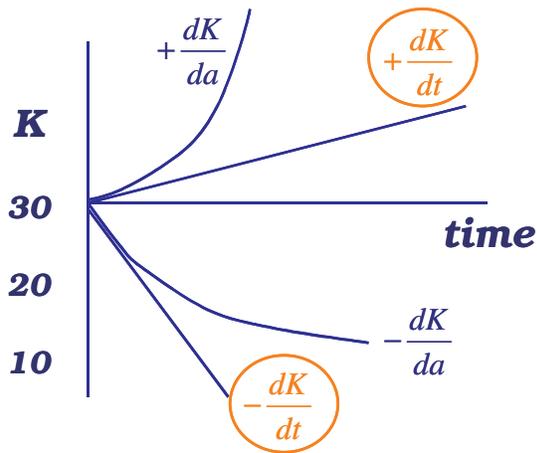


Figure B.1.11 Schematic and example crack length vs. time data showing the important role of  $+\frac{dK}{da}$  on crack growth rate, and the important difference between  $\frac{dK}{da}$  and  $\frac{dK}{dt}$  testing. In general,  $\frac{dK}{dt}$  always leads to non-conservative growth rates.  $+\frac{dK}{da}$  provide a positive feedback that often leads to dramatic increases in crack growth rate, while  $-\frac{dK}{da}$  represents negative feedback that slows the change in K as the crack slows down [24]. (Reprinted with Permission from ASME)

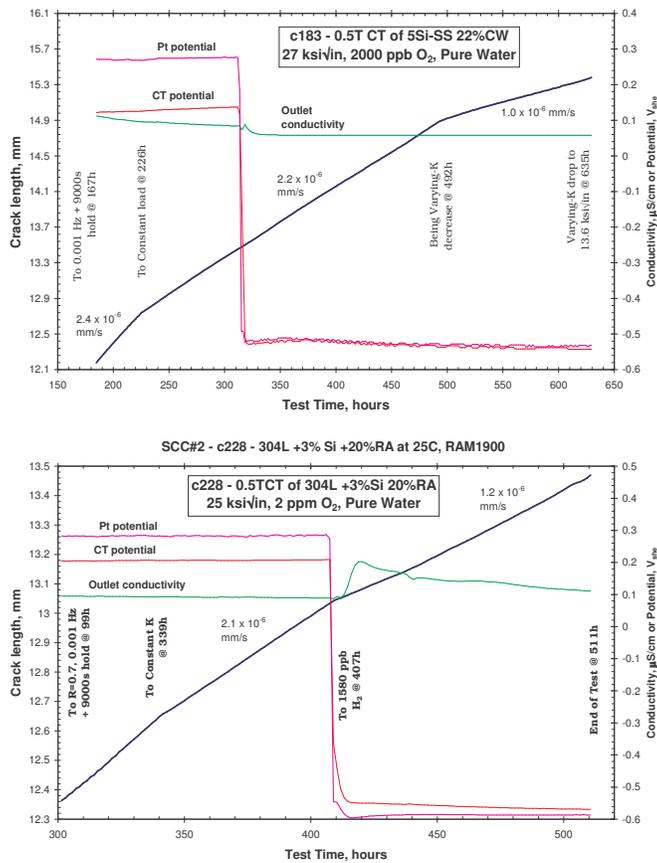


Figure B.1.12 Effect of Si on the crack growth rate of “stainless steels” whose bulk composition was designed to simulate an irradiated grain boundary. Reasonable questions can be raised about how realistic it is to create bulk alloys that represent the composition of a few nm region, but the qualitative observations remain important – i.e., that there is little effect of corrosion potential or stress intensity factor on crack growth rate [16,17,25]. (© 2003 by The American Nuclear Society, La Grange Park, Illinois)