An Industrial Perspective on Environmentally Assisted Cracking of Some Commercially Used Carbon Steels and Corrosion-Resistant Alloys

YUGO ASHIDA,^{1,3} YUZO DAIGO,^{2,4} and KATSUO SUGAHARA^{2,5}

1.—NHK International Corporation, 46855 Magellan Drive, Novi, MI 48377, USA. 2.—Hitachi Metals MMC Superalloy Ltd., 1230 Kamihideya, Okegawa-shi, Saitama Prefecture 363-8510, Japan. 3.—e-mail: yugo.ashida@nhkusa.com. 4.—e-mail: daigo@hitachi-metals-sa.co.jp. 5.—e-mail: sugahara@hitachi-metals-sa.co.jp

Commercial metals and alloys like carbon steels, stainless steels, and nickel-based super alloys frequently encounter the problem of environmentally assisted cracking (EAC) and resulting failure in engineering components. This article aims to provide a perspective on three critical industrial applications having EAC issues: (1) corrosion and cracking of carbon steels in automotive applications, (2) EAC of iron- and nickel-based alloys in salt production and processing, and (3) EAC of iron- and nickel-based alloys in supercritical water. The review focuses on current industrial-level understanding with respect to corrosion fatigue, hydrogen-assisted cracking, or stress corrosion cracking, as well as the dominant factors affecting crack initiation and propagation. Furthermore, some ongoing industrial studies and directions of future research are also discussed.

INTRODUCTION

Environmentally assisted cracking (EAC) is one of the most common causes of failure in structures and components. For decades, the mechanisms of EAC have been intensively and extensively studied in chemical processing, in the oil and gas industry, and in nuclear power generation.^{1,2} Having a better understanding of EAC is vital for product life design or life extension across industry sectors. This article discusses on the EAC problem of some commercially used metals and alloys. It is especially focused on low-cost carbon steels (CS) and low-alloy steels (LAS) used for the automotive industry, and corrosion-resistant Fe-Cr-Ni alloys, such as stainless steels and Ni-based alloys used in a salt production environment and in a supercritical water environment. The purpose of this review is to prevent or mitigate the risk of EAC by raising awareness of the failure mode and the susceptible regime in these industrial applications.

Table I shows the EAC problems of each industrial application described in this article.

The information includes materials, the working environment, the corrosion type/cracking mode, the current industrial-level understanding of the root

cause, controlling factors affecting the susceptibility to cracking, experimental approaches, and research efforts within each industry. In the following, we describe the EAC phenomena in the aforementioned industrial applications in three topical themes (categorized by application area).

CORROSION AND CRACKING OF CARBON STEELS IN AUTOMOTIVE APPLICATIONS

Low-cost carbon steels are used for automotive components, construction structures, bridges, and many other structures associated with daily life. A World Steel Association estimate shows that the automotive sector consumes roughly 12% of the overall steels in the world.³ On average, 900 kg of steel is used per vehicle, including 12% of high-strength carbon steels in the suspension.⁴ The corrosion of carbon steels caused by road salts and emissions was a serious problem during the early years of automotive development.⁵ In recent years, the corrosion and cracking problem was significantly improved through the advancement of coating technologies.⁵ Nevertheless, problems of localized corrosion, corrosion fatigue, and

Table I. Industrial application and environmentally assisted cracking of commercially used Fe-Cr-Ni alloys

Material	Carbon steel/low-alloy steel	Stainless steel	Ni-based alloy
Industry sector ^a	Automotive	Salt production/chemical processing	Salt Production/Super-critical Water
Chemical composition	Fe, C, minor elements	Fe, Cr > 10%	Ni ≥ 50%, Cr, Mo, etc.
Unit price ^b	<1 USD/kg	2–3 USD/kg	>10 USD/kg
Application	Such as automotive suspension coil springs	Corrosion and heat resistance	Outstanding corrosion and high-temperature resistance
Environment	Moist/humid air with [Cl ⁻], [SO ₄ ²⁻], etc.	High [Cl ⁻] concentrations	Supercritical water with [Cl ⁻], [SO ₄ ²⁻], etc.
Corrosion type	LC, CF, HAC	SCC in bulk material	LC, general corrosion, SCC
Root cause of failure	LC promoted CF and HAC from coating damaged locations	Low pH of concentrated [Cl ⁻]	CLSCC/SCC in SCW containing anions
Controlling factors	Substrate, surface modification, pretreatment, coating, coating damage	pH, [Cl ⁻]	Dielectric constant of water, species, and concentration of anions, electric potential
Experimental approaches	SST, CCT	U bend testing in high-temperature water containing chloride	Autoclave testing, slow strain rate testing
Required performance	Improved fatigue/corrosion fatigue life and resistant to delayed failure	Resistance to SCC	Excellent resistance to corrosion and SCC

LC localized corrosion, CF corrosion fatigue, HAC hydrogen assisted cracking, SCC stress corrosion cracking, SST salt spray test, CCT cyclic corrosion test.^aTypical industry sectors as described in this article.^bIndustrial metals prices and charts (<https://www.quandl.com/collections/markets/industrial-metals>, accessed on 10-28-2016).

hydrogen-assisted cracking in automotive components remain unsolved. The following section shows the typical case of automotive suspension coil springs.

Coating and Coating Damage Caused Corrosion and Cracking of Coil Springs

Carbon steels or low-alloy steels with pretreatment and resin powder coatings are being used for the production of coil springs, taking into account the balance between mechanical properties and the cost of materials and manufacturing. As shown in Fig. 1, the cross section of a coil spring wire consists of a metal substrate, a shot peened surface layer, a zinc phosphate (ZnP) pretreatment layer, and an organic coating layer. There is a transition zone of mechanical properties between the metal substrate and the shot peened layer. The chemical composition is not changed in the metal transition zone. Nevertheless, the interface between metal and ZnP and the interface between ZnP and the coating provide favorable sites for corrosion when the bonding condition is an issue, and a crevice is easily formed in a moist/humid environment. Leidheiser⁶ systematically summarized the problem of corrosion on coating protected metals and pointed out that

many problems remain unsolved. Ashida^{7,8} reported significant occurrence of localized corrosion from coating-damaged locations on coil springs.

It is important to understand localized corrosion under imperfect or damaged coatings. Localized corrosion has been studied for decades, and the current understanding of the mechanisms and critical factors has been reported.^{9–15} In the case of resin powder-coated carbon steels, susceptibility to corrosion is correlated to the properties of each protective layer, and the bonding conditions between different layers, as shown in Fig. 1. The main purpose of the shot-peened layer is to provide compressive stress on the surface to mitigate the initiation of fatigue cracks. In addition, the beneficial effect of compressive stress on localized corrosion is reported by Liu et al.¹⁶ and Peyre et al.¹⁷ The surface condition of shot-peened carbon steel affects the formation and adhesion of the ZnP layer and the organic coatings. Zinc phosphate crystals protect the metal surface from corrosion and provide a suitable interface for bonding with top coatings.¹⁸ The multiple chemical processes of forming the ZnP layer determines the composition/shape of particles, the surface coverage, and the layer thickness. The outermost layer of a single- or dual-powder coating provides a barrier on top, to prevent diffusion

Fig. 1. Schematic overview shows the process of corrosion and cracking originated from coating damage.

processes and the occurrence of chemical and electrochemical reactions on the metal surface. Imperfect or damaged coating can cause corrosion and subsequent crack initiation from the corroded site.

Figure 1 shows a schematic overview of corrosion and cracking that originated from coating damage. The process of localized corrosion (i.e., pitting or crevice corrosion) involves the formation of coating damage, partial or complete penetration through coating and pretreatment, dissolution of the exposed substrate (i.e., the carbon steel substrate corrodes when no protective layers are provided), crack initiation at the bottom of a corrosion pit, and hydrogen-assisted crack propagation that leads to final failure. In addition, crevice corrosion occurs when adhesion is lost at the interfaces. Kendig and Mansfeld^{19,20} reported that penetration of the coating did not by itself cause failure of the coated metal. Nevertheless, loss of adhesion between the coating and the surface, as well as cohesive failure, was the major driving force causing final failure.

Root Cause of Fatigue and Corrosion Fatigue

Fatigue cracking initiates in both dry and wet environments under cyclic loading. In most cases of field failure, the root cause of cracking is related to metallurgical, physical, or corrosion defects on the surface, such as surface or subsurface inclusions, mechanical scratches or improper shot peening, or pitting and possibly crevice corrosion. Dry fatigue commonly starts from a metallurgical or physical defect. In the case of corrosion fatigue, inclusions and physical defects (e.g., a damaged coating) also provide favorite sites for the initiation of corrosion

pits. Starting from these imperfect surfaces, small cracks form at the early stage of cracking. The formation of small cracks could represent a significant fraction of the life of a component.²¹ Subsequently, both aqueous environments and the mechanical driving force promote the propagation of corrosion fatigue.

Hydrogen-Assisted Cracking of High-Strength Carbon Steels

Hydrogen-delayed fracture or hydrogen-assisted cracking (HAC) of high- and ultrahigh-strength carbon steels have been studied for decades. According to Gangloff,²² hydrogen degradation of the crack propagation resistance of this type of alloy is categorized as either internal hydrogen-assisted cracking (IHAC) or hydrogen environment-assisted cracking (HEAC). As shown in Fig. 1, the effect of hydrogen is localized to the crack tip, regardless of the source of the hydrogen (i.e., introduced by manufacturing processes or through environmental exposure). In the case of suspension coil springs, the influences of tensile strength, hydrogen concentration, and applying stress mode (tension or torsion) on delayed fracture behavior have been experimentally investigated using SAE9254 steel with a tensile strength in the range between 1700 and 2000 MPa.²³ It was found that the delayed fracture strength tends to decrease with the increment of diffusible hydrogen concentration, which is especially significant when the amount of hydrogen is less than 1 ppm. To ensure design integrity, these findings on hydrogen cracking are important to design engineers.

Table II. U-bend test results in a boiling 45 wt.% MgCl₂ solution

UNS no.	Time until cracking in 45% MgCl ₂ at 154°C	Ni content, mass%	PRE number ^a
S30400	1–2	8.8	18
S31600	1–2	12.7	25
N08020	23	34.3	28
N08825	45	38.8	31
N06625	1000, no cracking	61.7	51
N10276	1000, no cracking	60.4	71
N06022	1000, no cracking	58.1	71
N06210	1000, no cracking	60.9	82

^aPRE = [Cr, wt.%] + 3.3([Mo, wt.%] + 0.5[W, wt.%]).

ENVIRONMENTALLY ASSISTED CRACKING OF IRON- AND NICKEL-BASED ALLOYS IN SALT PRODUCTION AND PROCESSING

Stress corrosion cracking of stainless steels in chloride containing environments is the most common failure mode of vessels and pipes in chemical processing and in the petrochemical industry. The mechanism and controlling factors affecting chloride stress corrosion cracking (CLSCC) have been systematically summarized in well-known corrosion textbooks.^{24,25} According to the empirical rule as shown in the Copson curve,²⁴ the susceptibility of CLSCC increases with the nickel content from 0% to approximately 10%. Nickel content from 8 to 10% is the typical range of traditional AISI 304 and 316 steels, that are in the range of maximum susceptibility. Nickel as an alloy element is said to have a beneficial effect on stress-corrosion cracking in chloride-laden environments in the case of nickel content being more than approximately 10%. The CLSCC susceptibility decreases for lower nickel content (i.e., duplex stainless steels) or higher content than this range, as in the case of super-austenitic stainless steels or nickel alloys. Alloys containing more than 42 wt.% nickel become immune to CLSCC. This article focuses particularly on the application of stainless steels and Ni-based alloys in extremely concentrated chloride environments.

In the Japanese salt manufacturing process, high-salinity water is obtained using an ion exchange membrane and distillation from sea water. The highest chloride concentration is around 35 wt.%,²⁶ and the highest process temperature is around 130°C.²⁷ Type 316L SS (S31600) is used in the standard structural material of salt manufacturing plants. Although there is extensive research about SCC in water containing chloride, only a few studies have been conducted under salt manufacturing conditions.²⁸

Tsugawa et al.²⁹ evaluated CLSCC susceptibilities of alloys S31600, S31254, S32053, N08354, and N06022 by U-bend tests under simulated salt manufacturing conditions. The alloys S31254, S32053, and N08354 are classified as super austenitic stainless steels, and alloy N06022 is classified as a nickel-

based, corrosion-resistant alloy. They reported that alloys with pitting resistance equivalent (PRE) numbers over 50, such as alloys N08354 and N06022, were proper structural materials for salt manufacturing plants with high chloride concentrations. The CLSCC susceptibilities of alloys S31600, S31254, and S32053, especially in a bittern are very high. The likely reason is that the high MgCl₂ proportion of the chlorides in a bittern brings about low pH and a low pH buffer capacity; that is, low pH makes passivation films of the alloys unstable, and the low pH buffer capacity forms a low pH region locally.

Sugahara³⁰ evaluated CLSCC susceptibilities of iron-based and nickel-based alloys by carrying out U-bend tests in a boiling 45 wt.% MgCl₂ solution as shown in Table II.

Although all of the alloys containing less than 39% nickel cracked within a short time, all of the alloys containing around 60% nickel did not crack even after being held for 1000 h. The results are in good agreement with the Copson curve.

ENVIRONMENTALLY ASSISTED CRACKING OF IRON- AND NICKEL-BASED ALLOYS IN THE APPLICATION OF SUPERCRITICAL WATER

The fourth state of water, supercritical water (SCW), is water at a higher pressure and a higher temperature than the critical point of water (374.0°C and 22.1 MPa). Supercritical water oxidation (SCWO),^{31–35} organic synthesis,^{36–44} hydrogen production and upgrading of heavy crude oil,^{45–48} and supercritical water reactor^{49,50} have been proposed as SCW applications. As SCW applications tend to be under extremely high corrosive conditions (e.g., high temperature and high pressure, high or low pH, oxidizing or reducing environment, existence of chloride ion, etc.), it is easy to cause the SCC problem. Rebak⁵¹ found that N06625 and N102276 are susceptible to intergranular stress corrosion cracking (IGSCC) under SCWO conditions by conducting slow strain rate testing (SSRT). In this article, the results of testing under conditions of

SCWO and reducing (including acidic and alkaline) are summarized on several more iron- and nickel-based alloys after Rebak's⁵¹ report.

Corrosion and SCC in SCWO

Harmful waste, such as chemical weapons and polychlorinated biphenyl (PCB), can be decomposed under SCWO conditions. It is possible to generate chloride and sulfuric ions in the process of decomposition. It is important to develop iron- and nickel-based alloys further with high performance of resistance to corrosion and SCC in SCWO. Accordingly, the effects of alloying elements on corrosion and cracking behavior of conventional and new alloys were investigated. Yakuwa et al. conducted corrosion tests of the Ni-(20–50 wt.%)Cr-(0–10 wt.%)Mo alloy in SCW containing several anions ($\text{Cl}^- = 17,000 \text{ mg/L}$, $\text{SO}_4^{2-} = 120 \text{ mg/L}$, etc.) and cations ($\text{Ca}^+ = 7,180 \text{ mg/L}$, $\text{K}^+ = 5,800 \text{ mg/L}$, etc.) at 600°C and 25 MPa. They reported that alloys with chromium content of more than 35 wt.% showed higher corrosion resistance than those with chromium content of less than 35 wt.%, and that molybdenum improved corrosion resistance of the alloys.⁵² Hara et al. conducted corrosion tests of nickel-based and iron-based alloys in SCW containing 0.001 mol/kg-HCl + 0.3 mol/kg- O_2 at 500°C 29.4 MPa. They reported that the corrosion resistance of tested alloys strongly depended on chromium content, and that nickel-based alloys with chromium content of more than 20 wt.% and iron-based alloys with chromium content of more than 30 wt.% became more corroded because of trans-passivation.⁵³ Schorer et al. conducted corrosion tests of Ni-(20–40 wt.%)Cr alloys in SCW containing 0.12 mol/kg-HCl + 0.06 mol/kg- O_2 at 410°C and 400 bar. They reported that Ni-30wt.%Cr showed excellent corrosion, and that tungsten improved the corrosion resistance of the alloys.⁵⁴ Watanabe et al. conducted corrosion tests on several types of nickel-based alloys and iron-based alloys in SCW containing sulfuric acid. They reported that chromium was a key element that improved the corrosion resistance of the alloys, and that iron also improved corrosion resistance of these alloys under oxidizing conditions.⁵⁵ They also reported that corrosion conditions of subcritical water and high-pressure SCW are more severe than lower pressure SCW because density and dielectric constant of water of those conditions are higher than those of lower pressure SCW. Higher density and the electric constant of water enhanced electrochemical reactions.⁵⁵ Fujisawa et al. conducted SSRT on S31603, N10276, N06625, and N06044 in SCW containing 0.001 and 0.01 mol/kg-HCl at 400°C 25 MPa. They reported that IGSCC susceptibility of tested alloys in 0.01 mol/kg-HCl was higher than in 0.001 mol/

kg-HCl. They also reported that the susceptibility to IGSCC-containing higher chromium showed lower IGSCC susceptibility.⁵⁶ In summary, the chromium content of the alloys is one of the key factors affecting corrosion and cracking behavior in SCWO.

In addition, Watanabe et al. evaluated SCC susceptibility of S31603 in the subcritical and SCW containing 0.01 mol/kg H_2SO_4 and reported that SCC susceptibility of S31603 was higher in SCW than in the subcritical.⁵⁷ In SCW, the sulfate concentration did not affect SCC susceptibility of S31603 in the range of 0.0001–0.01 mol/kg at 400°C and 25 MPa.⁵⁸ Nevertheless, it was found that a thick oxidized film formed along grain boundaries in a sulfate-containing environment. Water density was another factor affecting SCC susceptibility. Crack initiation in low-density water occurred more often than in high-density water.⁵⁸

Corrosion and SCC in Reducing (Acidic and Alkaline) SCW

Adding hydrogen to SCW (i.e., reducing SCW) can be used for refining coal and oil sand at high pH and for hydrothermal synthesis at low pH. Experimental data of SSRT on stainless steels and nickel-based alloys in reducing SCW were first reported by Nishida and Fujisawa et al.^{59,60} They evaluated the corrosion resistance of S31600, N10276, N06625, N06210, and N06044 under reducing 0.01-mol/kg-HCl and 0.05–0.5-mol/l-NaOH conditions.^{59,60} It was found that the corrosion resistance of nickel-based alloys showed better corrosion resistance over S31600 in all tested conditions, and that the corrosion rates of tested alloys tended to increase with increasing hydrogen partial pressure in SCW at low pH at 400°C and 25 MPa. They also showed that the Ni-Cr-Mo alloy demonstrated better corrosion resistance than a N06044 alloy that contains 44 wt.% of chromium. Fujisawa et al. also showed that alloy N06210 demonstrated the best corrosion resistance in those alloys, and that chromium and molybdenum improved corrosion resistance of these alloys under the tested conditions. At high pH, a similar corrosion behavior was observed. Regarding SCC susceptibility, Fujisawa et al.⁵⁶ found that the IGSCC susceptibility of S31600 and N10276 decreased when H_2 partial pressure was higher than 0.48 MPa. IGSCC susceptibility of S31600, N10276, N06625, N06210, and N06044 increased with increasing NaOH concentration at 400°C and 25 MPa. The SCC susceptibility of the tested alloy also decreased with increasing chromium content. In summary, as in SCWO, adding chromium to these alloys improved resistance to corrosion and SCC in reducing SCW. Molybdenum was also an effective element to increase corrosion and cracking resistance.

Fig. 2. SCC susceptibility maps of S31600 and N10276 in sub- and supercritical water: (a) SCC map of S31600⁶¹ and (b) SCC map of N10276.⁶¹

Mapping SCC in SCW

Figure 2 shows applicable conditions of iron- and nickel-based alloys in SCW. Figure 2a and b comprise the SCC maps of S31600 and N10276, respectively, which generated from experimental data. The SCC susceptible region is larger under oxidizing conditions than under reducing conditions for the two alloys. Both S31600 and N10276 are immune to SCC in the pH range between 3 and 11 in reducing SCW. The two alloys can be used for the

application of hydrogen production from biomass, which is in the pH range of 3.5–7.0 and the hydrogen partial pressure range of 0.01–3.00 MPa in reducing SCW.

INDUSTRY-DIRECTED RESEARCH AND FUTURE DIRECTIONS

For industrial applications, it is essential to identify root causes and key factors affecting the initiation and propagation of EAC. This article

shows a typical range of variation in material-environment combination. In the case of a high corrosion rate ($\geq 10 \mu\text{A cm}^{-2}$) material, such as carbon steel, exposure to a mild aqueous environment can cause localized corrosion to accelerate the initiation of fatigue or corrosion fatigue significantly. For corrosion-resistant alloys with passivation behavior, the PRE number works as an indicator of the passive film broken down. Increasing the number of PRE delays the occurrence of pitting and effectively prevents SCC starting from pitting corrosion. In addition, mapping of a potential SCC susceptible region becomes more important for proactive design under extreme environments such as SCWO and reducing SCW. In general, to mitigate EAC in commercially used materials and structures, the research from industry includes developing new steels and alloys to improve EAC resistance, developing new testing technologies to identify the EAC regime, and increasing integrity and reliability to extend the life of components and structures. The future direction of industry-directed research will need to continuously focus on these areas to meet the requirements of next-generation production and applications.

Regarding new steel grades for coil spring manufacturing, Nakayama et al.^{62–65} proposed a method to improve corrosion fatigue resistance of ultrahigh-strength steel by rust control and hydrogen trapping. In the development of UHS1900 steel designed with high strength and excellent corrosion fatigue resistance, rust composition control was achieved by the addition of copper, nickel, and chromium. Hydrogen trapping sites were introduced by the addition of titanium. In addition, Yoneguchi et al.⁶⁶ reported the effectiveness of adding boron, niobium, and nickel for the improvement of corrosion fatigue strength and delayed fracture strength of coil spring steels. The development of the N06210 alloy previously described for application in highly concentrated chloride solutions and SCW environments is another example of improving EAC resistance. It's important to develop iron- and nickel-based alloys further with a high performance of resistance to corrosion and SCC in the SCW environment.

Industries are interested in developing new methodologies for testing corrosion and cracking. For automotive applications, Ashida et al.⁶⁷ showed the effectiveness of using coupled multielectrode array sensor (CMAS) technology to evaluate localized corrosion on ZnP pretreatment and powder coatings. Otake et al.⁶⁸ used a compact neutron source for the nondestructive visualization of corrosion under coating films. The new technologies help to better understand the mechanism of corrosion and cracking under coatings, or at a damaged coating site. For testing in the SCW environment, it is important to enhance the capability of testing EAC, as well as to generate reliable data in the environment up to 30 MPa of pressure and 600°C.^{31–50}

To achieve both high performance and extended component life for industrial applications, it is crucial to evaluate EAC resistance along with material development, component design, and manufacture processing. The root cause of failure and the key factors affecting the engineering life of components must be identified by testing not only small specimens but also component-level performance. These concepts apply in the development of 2000-MPa grade, advanced high-strength spring steel for the suspension component of lightweight vehicles.^{66,69} In the meantime, surface condition control before coating (including the control of soluble salt contamination⁷⁰) is important for the management of the coating life cycle, which largely affects the initiation of localized corrosion and EAC in high-strength carbon and low-alloy steels.

ACKNOWLEDGEMENTS

The authors would like to thank Dr. Shinichi Nishizawa of NHK International Corporation and Mr. Toru Kobayashi of Hitachi Metals MMC Superalloy, Ltd. for their support in the preparation of this article. We thank Dr. Srujan Rokkam, the SCC Subject Editor, and the four anonymous reviewers for their comments.

REFERENCES

1. G.S. Was, Y. Ashida, and P.L. Andresen, *Corros. Rev.* 29, 7 (2011).
2. R.B. Rebak, in *Stress Corrosion Cracking – Theory and Practice*, ed. by V.S. Raja, T. Shoji (Woodhead Publishing, Cambridge, 2016), p. 273.
3. Trefis Team, Trends in steel usage in the automotive industry (Forbes.com, May 20, 2015), <http://www.forbes.com/sites/greatspeculations/2015/05/20/trends-in-steel-usage-in-the-automotive-industry>. Accessed September 30, 2016.
4. World Steel Association, Automotive (Website of the World Steel Association, 2016), <https://www.worldsteel.org/Steel-markets/Automotive.html>. Accessed September 30, 2016.
5. U.S. Department of Transportation, Report No. FHWA-RD-01-156, Federal Highway Administration, Washington, DC (2002).
6. H. Leidheiser Jr., *Corrosion* 38, 374 (1982).
7. Y. Ashida, *CORROSION 2016* (NACE International, Houston, TX, 2016), Paper No. 7287.
8. Y. Ashida, *CORROSION 2017* (NACE International, Houston, TX, 2017), Paper No. 9152.
9. G.S. Frankel and N. Sridhar, *Mater. Today* 11, 38 (2008).
10. G.S. Frankel, *J. Electrochem. Soc.* 145, 2186 (1998).
11. A. Turnbull, *Corros. Sci.* 23, 833 (1983).
12. C. Gabrielli, F. Huet, M. Keddam, and R. Oltra, *Corrosion* 46, 266 (1990).
13. C. Gabrielli and M. Keddam, *Corrosion* 48, 794 (1992).
14. H.-H. Strehblow, *Werkst. Korros.* 35, 437 (1984).
15. S.M. Sharland, *Corros. Sci.* 27, 289 (1987).
16. X. Liu and G.S. Frankel, *Corros. Sci.* 48, 3309 (2006).
17. P. Peyre, X. Scherpereel, L. Berthe, C. Carboni, R. Fabbro, G. Beranger, and C. Lemaitre, *Mater. Sci. Eng. A* 280, 294 (2000).
18. H.S. Bender, G.D. Cheever, and J.J. Wojtkowiak, *Prog. Org. Coat.* 8, 241 (1980).
19. M. Kendig, S. Tsai, and F. Mansfeld, *Mater. Perform.* 23, 37 (1984).
20. F. Mansfeld and M. Kendig, Report No. AD-A167281, Metallurgy & Ceramics Program, Material Sciences Division, Office of Naval Research, Arlington, VA (1986).

21. A. Turnbull, *Proc. R. Soc. A* 470, 20140254 (2014).
22. R.P. Gangloff, in *Environment Induced Cracking of Materials (EICM-2)*, ed. by S Shipilov, R.H. Jones, J.-M Olive, R.B. Rebak (Elsevier Science, Oxford, UK, 2007), p. 141.
23. Research Committee on Delayed Fracture of Spring Steels, *Bane Ronbunshu* 58, 41 (2013).
24. H.R. Copson, *Effect of Composition on Stress Corrosion Cracking of Some Alloys Containing Nickel, Physical Metallurgy of Stress Corrosion Fracture* (New York: Interscience Publishers, 1959).
25. H.S. Khatak and R. Baldev, *Corrosion of Austenitic Stainless Steels: Mechanism, Mitigation and Monitoring* (Materials Park: ASM International, 2002).
26. F. Yamada, M. Hidaka, T. Kaneko, and T. Fukumoto, *Bull. Soc. Sea Water Sci. Jpn.* 61, 141 (2007).
27. Y. Nagaoka and S. Aida, *Bull. Soc. Sea Water Sci. Jpn.* 27, 272 (1974).
28. N. Shinohara, *Boshoku-Gijutsu* 29, 497 (1980).
29. T. Tsugawa, Y. Watanabe, and H. Abe, *Bull. Soc. Sea Water Sci. Jpn* 65, 88 (2011).
30. K. Sugahara, *J. Soc. Mater. Sci. Jpn* 64, 57 (2015).
31. M. Modell, U.S. Patent, No. 4,338,199 (1982).
32. E.F. Gloyna and L. Li, *Waste Manag.* 13, 379 (1993).
33. Y. Arai, S. Koda, K. Nagahama, T. Sako, and K. Totigi, *Fundamentals and Applications of Supercritical Fluids* (Tokyo: Technosystem Co., Ltd., 2002).
34. Y. Akai, in *International Symposium on Innovative Reaction Fields, Proceedings* (2004), p. 173.
35. S. Baur, H. Schmidt, A. Kramer, and J. Gerber, *J. Supercrit. Fluids* 33, 149 (2005).
36. Y. Ikushima, *Chem. Eng.* 48, 173 (2003).
37. T. Sato, G. Sekiguchi, T. Adschiri, and K. Arai, *Chem. Commun.* 17, 1566 (2001).
38. T. Sato, G. Sekiguchi, T. Adschiri, and K. Arai, *Ind. Eng. Chem. Res.* 41, 3064 (2002).
39. T. Adschiri, R. Shibata, and K. Arai, *J. Jpn. Petrol. Inst.* 40, 291 (2002).
40. D.T. Chen, C.A. Perman, M.E. Riechert, and J. Hoven, *J. Hazard. Mater.* 44, 53 (1995).
41. T. Adschiri, O. Sato, K. Saito, and K. Arai, *Kagaku Kogaku Ronbunshu* 23, 505 (1997).
42. M. Sasaki, Z. Fang, T. Fukushima, T. Adschiri, and K. Arai, *J. Ind. Eng. Chem. Res.* 39, 2883 (2000).
43. T. Adschiri, S. Nagashima, H. Shibuichi, K. Arai, and M. Shishido, *J. Jpn Inst. Energy* 75, 742 (1996).
44. Y. Hakuta, T. Adschiri, and K. Arai, *Chem. Eng.* 45, 45 (2000).
45. D.S. Ross, J.E. Blessing, Q.Q. Nguyen, and G.P. Hum, *Fuel* 63, 1206 (1984).
46. H. Enomoto and M. Kinoshita, *J. Inst. Saf. High Press. Gas Eng.* 40, 14 (2003).
47. D. Yu, M. Aihara, and M.J. Antal Jr., *Energy Fuels* 7, 574 (1993).
48. M.J. Antal Jr., S.G. Allen, D. Schulman, X. Xu, and D. Robert Jr., *Ind. Eng. Chem. Res.* 39, 4040 (2000).
49. U.S. DOE Nuclear Energy Research Advisory Committee and the Generation IV International Forum, A Technology Roadmap for Generation IV Nuclear Energy Systems, Report No. GIF002-00, 1 December (2002) (<http://nuclear.gov>).
50. P. MacDonald, J. Buongiorno, C. Davis, and K. Weaver, Feasibility study of supercritical light water cooled reactors for electric power production. Final Report, DEFG07-02SF22533, INEEL/EXT-04-02539, January (2005). (<http://wins.engr.wisc.edu/research/scw/intl/SCIEM-II/INEEL.pdf>).
51. R.B. Rebak, *CORROSION 2005* (NACE International, Houston, TX, 2005), Paper No. 05457.
52. H. Yakuwa, M. Miyasaka, and S. Imura, *CORROSION 2002* (NACE International, Houston, TX, 2002), Paper No. 02361.
53. N. Hara, S. Tanaka, S. Soma, and K. Sugimura, *CORROSION 2002* (NACE International, Houston, TX, 2002), Paper No. 02358.
54. C. Schroer, J. Konys, J. Novotny, and J. Hausselt, *CORROSION 2002* (NACE International, Houston, TX, 2002), Paper No. 02359.
55. Y. Watanabe, K. Shoji, T. Adchiri, and K. Sue, *CORROSION 2002* (NACE International, Houston, TX, 2002), Paper No. 02355.
56. R. Fujisawa, K. Nishimura, T. Nishida, M. Sakaiharu, Y. Kurata, and Y. Watanabe, *CORROSION 2005* (NACE International, Houston, TX, 2005), Paper No. 05395.
57. Y. Watanabe, H. Abe, and Y. Daigo, *CORROSION 2004* (NACE International, Houston, TX, 2004), Paper No. 04493.
58. H. Abe, Y. Daigo, Y. Watanabe, R. Fujisawa, and M. Sakaiharu, in *Proceedings of the 50th Japan Conference on Materials and Environments* (Japan Society of Corrosion Engineering, Okinawa, Japan, 2003), D-115, p. 365.
59. R. Fujisawa, M. Sakaiharu, Y. Kurata, and Y. Watanabe, *Corros. Eng. Sci. Technol.* 40, 244 (2005).
60. R. Fujisawa, K. Nishimura, T. Nishida, M. Sakaiharu, Y. Kurata, and Y. Watanabe, *CORROSION 2005* (NACE International, Houston, TX, 2005), Paper No. 05392.
61. New Energy and Industrial Technology Development Organization, Research and Development on Environmental Burden Reduction Technologies with Supercritical Fluids Report (2006), (<http://www.nedo.go.jp/content/100096392.pdf>).
62. T. Nakayama and H. Tanaka, *Tetsu-to-Hagane* 100, 1089 (2014).
63. T. Nakayama, T. Iwata, M. Nagao, M. Shimotusa, N. Ibaraki, and A. Inada, *R&D Kobe Steel Eng. Rep.* 47, 50 (1997).
64. T. Nakayama, *Zairyo-to-Kankyo* 48, 484 (1999).
65. M. Shimotsusa, N. Ibaraki, T. Ikeda, and T. Nakayama, *Wire J. Int.* 31, 78 (1998).
66. A. Yoneguchi, J. Schaad, Y. Kurebayashi, and Y. Ito, *SAE 2000 World Congress* (SAE International, Warrendale, PA, 2000), Paper 2000-01-0098.
67. Y. Ashida, *CORROSION 2017* (NACE International, Houston, TX, 2017), Paper No. 9011.
68. Y. Otake, M. Yamada, A. Taketani, H. Sunaga, and T. Nakayama, *Isotope News* 722, 8 (2014).
69. W. Hui, H. Dong, Y. Weng, J. Shi, and M. Wang, in *Advanced Steels*, ed. by Y. Weng, H. Dong, Y. Gan (Springer, Berlin, Germany, 2011), p. 137.
70. H. Peters, *Mater. Perform.* 45, 30 (2006).