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Survey of the Degradation Modes of Candidate Materials for High-Level Radioactive Waste Disposal Containers

Iron-Base, Corrosion-Allowance Materials

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I. Introduction

The estimation of long-term container performance in a deep geologic repository environment, such as the potential site at Yucca Mountain, requires the evaluation of relevant materials degradation data. These data are required for the completion of performance assessment modeling efforts. Previous degradation modes surveys (Farmer, et al., 1988) have been completed for the six candidate materials from the Yucca Mountain Site Characterization Plan (DOE, 1988). An additional review of the degradation of nickel-base alloys was also completed by Gdowski (1991). These surveys have been employed to identify relevant failure mechanisms for the container materials under repository relevant conditions (Farmer, et al., 1991) and to provide a relative ranking of potential container materials based upon currently available data (Van Konynenburg, et al., 1993).

The evolution of the waste package design effort has resulted in an increased emphasis on the performance of the engineered barrier system (EBS). The EBS includes the waste package and the near-field engineered repository environment. This evolution has prompted research and analysis related to the development of a multiple barrier Advanced Conceptual Design (ACD) for the waste package disposal container. Table I-A shows the typical dimensions for a two-barrier ACD container.

The ACD systems currently under consideration consist of a multiple barrier design. The outer barrier will be fabricated from a corrosion allowance material, such as carbon steel. The inner barrier will employ a corrosion resistant material, such as Alloy 825. The thickness of each barrier is currently a variable in the container design process. The predicted long-term performance of each barrier material under repository relevant conditions will dictate the ultimate thicknesses of each barrier. Degradation modes surveys for Alloy 825 have been completed previously (Farmer, et al., 1988 and Gdowski, 1991). The purpose of the literature review effort undertaken in this report is the completion of a degradation modes survey for potential corrosion allowance materials that may be applicable to the ACD systems in the Yucca Mountain environment.

One of the most significant factors impacting the performance of materials under repository relevant conditions is the thermal environment. This environment will be affected by the areal power density of the repository, which is dictated by facility design, and the dominant heat transfer mechanism at the site. A significant factor related to repository thermal performance that impacts the degradation of the container is the time period over which each container is predicted to remain dry. During this time period, a majority of the degradation of the container will occur due to simple moist air oxidation.

Since more aggressive corrosion mechanisms become activated in wet environments, it is important to estimate the fraction of containers that remain dry as a function of time.

Following this dry-out period, the dominant degradation mechanism for the corrosion allowance materials will most likely be aqueous corrosion. This study presents a detailed review of the technical literature for the degradation modes of potential corrosion allowance candidate materials under repository relevant conditions.

II. Fundamentals of Corrosion

The following is a summary of the review of general corrosion by Scully (1966). Corrosion, or its scientific synonym, "oxidation" is an electrochemical reaction between two elements by which a metal loses an electron to the other species taking part in the reaction. This other species, oxygen, is said to have been reduced.

A metal oxide consists primarily of positively charged metal ions, Me^{z+}, where z is an integer, and larger, negatively charged oxygen ions, O^{2-}. Since the sum of the positive charges is equal to the sum of the negative charges, the overall oxide film is electrically neutral, although small regions can be either n-type or p-type semiconductors.

Oxides are predominantly ionic in character, although there does exist some covalent bonding. There is a continuous transition from ionic to covalent bonding. Where a compound lies between the two types depends upon the difference between the electronegativity of atoms involved in the process. The electronegativity of an atom describes the tendency of that atom to ionize.

Oxides often, but not always, behave similarly to metals. Oxides are composed of grains and exhibit grain growth. Oxide may recrystallize and deform plastically. In oxide films, as in metals, the grain growth, recrystallization, and deformation, being thermally activated processes, all increase at higher temperatures. Oxide crystals show higher diffusion rates along their intergranular interfaces, particularly at lower temperatures. Diffusion rates on the surfaces of oxide are higher than within the bulk of the oxide film.

II.A. Thermodynamics of Oxidation

Oxidation of a metal results in a change in the free energy of the system. It is this change in free energy of the system that drives the electrochemical reactions involved in oxidation of a metal. The change in free energy of the system is equal to the amount of work done or absorbed during the process minus the change in entropy of the system

($\Delta G=\Delta H-T\Delta S$). In order for the oxidation reaction to occur at all, the reaction must cause a reduction in the free energy of the system. The standard free energy change for nearly all oxides is negative. This means that oxides are thermodynamically stable in oxygen atmospheres whereas metals are not.

Since free energy changes are negative for the formation of metal oxides, all metals should revert to a combined, and therefore more stable, state when exposed to the atmosphere. It is due to the kinetics of oxidation reactions that this does not occur in times significant for the performance of engineering structures. The formation of a metal oxide on a bare metal surface usually restricts further access of one reactant to the other. When oxide thicknesses reach a certain value, 1-4 nm, oxide growth ceases. The oxide layer forms a protective barrier between the reacting metal and oxygen due to the continuous, non-porous layer that usually forms. In the event that the generated oxide layer is not continuous or is porous, the oxidation reaction continues and the metal will be converted to oxide. Thus, the very existence of metals and alloys is dependent on the ability of these materials to form protective oxide layers.

II.B. Initial Stages of Oxidation

Oxidation begins by a physical adsorption of oxygen molecules on the bare metal surface. These oxygen molecules then dissociate into atoms. At this point, oxygen atoms are dissolved in the metal. The atoms on the surface become much more strongly bound by a process of chemisorption that occurs with a much higher energy change, as high as 600 kJ/mol. This is up to 30 times as high as the energy of the adsorption process. There is experimental evidence that oxygen chemisorption is associated with the movement of a specific number of metal atoms into the plane of the adsorbed oxygen atoms. Together, the metal atoms and the adsorbed oxygen atoms form a very stable structure consisting of both positive and negative species that have been shown to be more stable than the bulk oxide. The transition of the monolayer to crystalline oxide must be explained by the effect of a second outer layer of chemisorbed oxygen molecules in altering the free energy balance so that the oxide becomes more stable than the monolayer.

II.C. Oxidation Kinetics

Oxidation rates for low temperatures are usually described with reference to the mathematical relationship between oxide thickness and time by the following formulas.

$y = k_1 * \log(t)$	logarithmic
$1/y = k_2 - k_3 * \log(t)$	inverse logarithmic
$y = k_4 * (1 - \exp(-k_5 * t)$	asymptotic

for thin films, and

$y^2 = k_6 * t$	parabolic
$y = k_7$	rectilinear

for thicker films. The terms, k_i, are constants, and the oxide thickness, y, is assumed to be uniform.

For very thin films, there is a strong electric field across the oxide layer that causes metal ions to cross through the film. As the film gets thicker [1-4 nm], the electric field is no longer sufficient to maintain oxide growth. Unless there is sufficient thermal energy present to cause continued growth by ionic diffusion under an activity gradient in the film, the oxidation rate will fall off rapidly. As long as oxide growth is dependent on the existence of electric fields, the oxide will never grow very much, even as oxidation temperature is increased.

The development of stresses in growing oxides may result in scale lift-off followed by the development of porosity. If lift-off is followed by scale fracture, then the newly exposed metal will exhibit very rapid oxidation.

II.D. Electrical Conductivity of Oxides.

Oxide growth can be compared with a current flow around a circuit containing an electrolytic cell. Oxide growth is characterized by the ionization of metal and oxygen atoms. These are anodic and cathodic processes, respectively. While the metal ions, or cations, are attracted toward the cathodic surface, the oxygen ions, or anions, are attracted to the anodic surface. During oxide growth there may be migration of both cations and anions or only one species may migrate while the other is immobile.

While oxidation occurs by ionic transport, there must be electron movement toward the cathode. This movement is largely determined by the activity gradient, in accordance with Fick's laws.

II.E. Oxidation of Iron and Iron Alloys

The oxidation rate of iron is sensitive to which crystal face is exposed to oxygen in the low temperature region (Ca. 250°C), due to the dependence of surface atom arrangement on the crystal face. Thus, the free energy of a solid surface varies with the crystal face. The oxidation rate decreases in the crystal face order (100) > (111) > (110). The oxide nuclei consist of Fe_3O_4, which then grows to form a uniform film of oxide. Next, α-Fe_2O_3 nucleates and covers the Fe_3O_4 layer (Uhlig and Revie, 1985).

Oxidation of iron leads to the generation of up to three distinct layers of iron oxide (Fe_2O_3, FeO, and Fe_3O_4) . The proportions of these layers change as the temperature and oxygen-partial-pressure changes.

Uhlig and Revie (1985) cite chromium and aluminum as the most efficient alloying elements for improving oxidation resistance of iron in air. Good oxidation resistance, combined with acceptable mechanical properties and ease of fabrication, account for the wide commercial application of chromium-containing iron alloys. The aluminum-containing iron alloys, however, exhibit poor mechanical properties, sensitivity of their protective oxide scales to damage, and a tendency to form aluminum nitride, all of which causes embrittlement of the alloy.

II.F. Atmospheric Corrosion

The following is a brief review of the fundamental aspects of atmospheric corrosion compiled by Swandby (1963). Since the atmosphere varies in minor constituents that affect its corrosive properties, the corrosion rates in a particular atmosphere may be several times those in another. Regardless of location, the atmosphere contains general corrosives, such as oxygen, carbon dioxide, and water. The amount of oxygen in air does not vary significantly, and the amount present is much greater than that portion actually taking part in corrosion reactions. The amount of carbon dioxide does vary with climatic conditions but is only present in small quantities. The amount of water varies greatly with location and is responsible for the greatest damage to metallic structures subjected to atmospheric corrosion.

Although water in air is the biggest problem from the corrosion standpoint, the effect of water on corrosion rates can be either beneficial or detrimental. Alloys that are directly exposed to the atmosphere tend to benefit from the rinsing action of rain, which rinses away some of the corrosives from the surface of an alloy. Water becomes a contributing factor to corrosion when a metal surface remains wet since the water tends to act as an electrolyte promoting the flow of electric current. In the absence of moisture on the metal surface, the corrosion of a metal is driven by gaseous oxidation reactions. Therefore, the relative humidity of an atmosphere becomes an important factor.

Many tests' results support the idea that corrosion rates increase with increasing relative humidity. Several experimenters have identified a "critical humidity", above which the corrosion process proceeds rapidly. This critical humidity varies for different metals and shows a dependence on the corrosion product film that forms on the metal.

In addition to the three general corrosives just mentioned, there may be other corrosives present in the air. Considering the presence of these other corrosives, we can classify an atmosphere as industrial, marine, or rural.

Industrial atmospheres contain soot, fly ash, and sulfur compounds. Sulfur compounds are more prevalent and active than any other impurity in the air. These compounds originate from the combustion of sulfur-containing fuels. They are ordinarily in a reduced state, but are partially oxidized by air, forming sulfur dioxide, and form sulfuric acid in the presence of moisture. The interaction of sulfur dioxide with iron and other metals may be considered to occur in 3 steps: sulfur dioxide adsorption, sulfur dioxide oxidation, and metal corrosion (Brown and Masters, 1982). Metal sulfates form on metals by exposures to atmospheres containing sulfur dioxide and sulfuric acid.

Marine atmospheres contain chlorides in the air. The chlorides originate from sea spray that is picked up by the wind. The small droplets of sea salt in the air become smaller by evaporation, forming microscopic particles of dry sea salt or salt-saturated water. Because of their hygroscopic nature, the salt particles are quite corrosive to many metals and alloys.

Rural atmospheres are not as corrosive as industrial and marine environments. In the absence of sulfates and chlorides, metals tend to corrode by the general factors previously discussed, such as moisture content.

Because of the large amounts of sulfates present in industrial areas, the industrial atmospheres are the most corrosive. On a mole-for-mole basis the chlorides present in marine atmospheres are more corrosive than sulfates. However, the concentrations of the chlorides are low, and the chlorides are more soluble in water than sulfates. Seacoast cities exhibit a marine atmosphere that also contains sulfates.

The factors that affect the corrosion rates of metals and alloys are temperature, relative humidity, rainfall, time-of-wetness, and the extent of pollution in the atmosphere (sulfur compounds, chlorides, and suspended particulates).

The effect of temperature on corrosion rate, according to a data survey compiled by Nelson (1967), depends on its influence on the factors controlling the corrosion reaction and the electrochemical potential of the metal. Nelson states that temperature definitely affects oxygen solubility and availability, the diffusion rate of oxygen, pH, and may effect the corrosion product film solubility and degree of protectiveness of the corrosion product film. Temperature may also affect the relative humidity and the time-of-wetness of a metal's corrosion product film. Relative humidity, rainfall, and time-of-wetness are interrelated. Rainfall may be beneficial to the extent that it may rinse away corrosives from the surface of the corrosion product film. Heavy rainfall, however, may encourage the flaking of scales from the corrosion product film, exposing bare surface metal that will corrode. Relative humidity is important because higher relative humidity increases the time-of-wetness of the corrosion product film leading to greater corrosion rates. In the absence of condensation, even in the most polluted atmospheres, the corrosion process is an electrochemical process due to the presence of a thin moisture film which can act as an electrolyte (Brown and Masters, 1982). Since corrosion is an electrochemical process, it will not occur when steel is not wet (Haynie and Upham, 1974). At high relative humidities (greater than about 70%), a continuous film of moisture forms on the metal surfaces. This film thickens with increasing relative humidity, leading to the electrochemical process of corrosion.

II.G. Summary of Oxidation and Atmospheric Corrosion Experiments

A number of experiments have been performed to find the correlations between corrosion rates and the factors previously mentioned. The purpose of this section is to present the experiments performed and to summarize the conclusions drawn from these experiments.

In an experiment performed by Haynie (1986), weathering steel samples were exposed for periods up to 30 months at nine air monitoring sites in the St. Louis, Missouri area. During exposure, temperature, dew-point, wind speed, wind direction, total oxidant, NO_x, SO_x, and total suspended particulate matter were monitored. Haynie used 153 sets of triplicate weight loss data along with environmental data to generate an empirical relationship between a specimen's weight loss and the environmental factors during exposure. This study concludes that three variables increase the diffusivity through the oxide film. These variables include 1) fraction of time wet, 2) temperature, and 3) total suspended particulate flux. The stability of the oxide film on weathering steels is decreased with increasing sulfur oxide concentration during periods of wetness. Dissolution of the oxide films on weathering steels is reduced by increasing the concentration of nitrogen oxides and increasing the water contact with the surface due to increased precipitation. Ozone appears to have no significant effect on the corrosion of weathering steel.

Vernon (1949) graphically shows the effects of relative humidity in conjunction with 0.01 percent sulfur dioxide, 0.01 percent sulfur dioxide and particles of ammonium sulphate, and 0.01 percent sulfur dioxide and particles of charcoal. Figure II-1 shows the effects of these contaminants, compared to pure air, on the atmospheric corrosion of mild steel with changing relative humidity. Vernon found that increases in relative humidity have no sensible effect of the corrosion rate in the absence of pollutants. Vernon's plots support the idea of a critical humidity, above which corrosion proceeds at an increased rate. At around 80 percent relative humidity, there is a marked increase in the corrosion rate of the steel sample when pollutants are present. Vernon concludes that at high relative humidities the controlling factor of corrosion is provided entirely by the state of purity of the atmosphere. In support of Vernon's conclusions are those of Aziz and Godard (1959), in that the complete absence of water vapor, sulfur dioxide has no influence on the normal room temperature oxidation that takes place in dry, pure air. It is necessary to have moisture for the electrochemical corrosion process to proceed.

Haynie and Upham (1974) exposed an enameling steel at 57 selected sites. This steel, with nominal 0.019 percent carbon and 0.028 percent copper, was selected for its susceptibility to atmospheric corrosion. The sites were chosen to represent normal and extremes of climatic and pollution conditions. During exposure, data was collected for average temperature, relative humidity, total suspended particulates, the amounts of sulfate and nitrate in the particulates (at over 90 percent of the sites), and sulfur dioxide (at 27 sites). The following conclusions were obtained: 1) total suspended particulates, nitrates in particulates, and the average annual temperature are relatively unimportant factors in the corrosion behavior of steel, 2) either sulfate in suspended particulates, atmospheric sulfur dioxide, or both, accelerate steel corrosion, 3) corrosion of steel should not be a problem in arid regions where the average relative humidity does not exceed 30 percent, and 4) savings from reduced corrosion of steel maintenance can be an appreciable benefit of reduced ambient levels of sulfur dioxide.

Knotkova-Cermakova, Vlckova, and Honzak (1982) completed a study on the atmospheric corrosion of weathering steels. The materials evaluated in this study included the low-alloy weathering steel, Atmofix, and a standard structural carbon steel (as a reference steel). The compositions of these materials are summarized in Table II-A. Each material was exposed to different atmospheric conditions, both indoors and outdoors. The corrosion for each material set of atmospheric conditions was measured as a function of time. The results of this experiment are summarized in Figure II-2. In rural and urban atmospheres with low levels of pollution (sulfur dioxide levels of about 40 mg/m^2/day or less) the conditions are suitable for the formation of a highly protective rust layer on low-alloy steels, resulting in a relatively stable low rate of corrosion. The development of this low stable corrosion rate occurs after about three years of exposure. The reference steels show a higher initial corrosion rate and maintain a higher corrosion rate. The reference carbon steels also take longer to reach their stable corrosion rate.

In more polluted atmospheres, when the annual average pollution level reaches 90 mg/m^2/day, the increased sulfur dioxide level causes a significant increase in the corrosion loss envelope curves, as shown in Figure II-2. The corrosion process of all steels initiates more rapidly in this intermediate pollution environment than in the low pollution environment, and the envelope curves begin to widen. The final corrosion rates are larger than for the low pollution environments. The low stable corrosion rate for the weathering steel takes longer to develop in the intermediate pollution environments, typically three to five years; but the steady-state values are still very low. The reference steel envelope curves show a fan shaped pattern. These steels are apparently not capable of providing protection against the higher sulfate concentrations resulting from higher sulfur dioxide contents. The stable corrosion rates of the carbon steels are significantly higher than those of the weathering steels.

For heavily polluted industrial atmospheres when the sulfur dioxide level is high, data indicates that the process of spalling and regrowth of the stable rust layer is occurring on both types of steel. This process appears to occur at regular interval and results in significant mass loss in the carbon steels. The performance of the weathering steels is clearly better than the carbon steels.

For corrosion behavior data of the steels in shed and indoor atmospheres, Knotkova-Cermakova, Vlckova, and Honzak (1982) made the following conclusions. Zero corrosion rates were not observed for either type of steel. There were not significant differences between the performance of the two types of steels. The results of the indoor exposures in both rural and urban atmospheres show no performance differences between the two types of steels. The protective rust layer that normally forms on weathering steels directly exposed to the atmosphere does not appear to be as protective in shed exposures.

Townsend and Zoccola (1982) analyzed the eight-year exposure data for weathering steel and steels with 0.021 percent and 0.21 percent copper in a marine, a rural, and two industrial environments. The experimenters conclude that A588 Grade B weathering steel is six to nineteen times more durable than the steel with 0.021 percent copper and two to ten times more durable than the steel containing 0.21 percent copper.

In an experiment by McKenzie (1982), Cor-Ten B, a weathering steel, and a special mild steel were exposed vertically, both under bridges and boldly exposed (chemical compositions are given in Table II-B). The specimens were exposed in five rural areas, three areas exposed to light industry, two marine areas, an industrial area, and a heavily industrial area. The concentrations of sulfur compounds and chlorides, temperature, relative humidity, and rainfall quantity at each of the sites were monitored.

McKenzie's conclusions are summarized as follows. Cor-Ten B generally corroded at a lower rate than the special mild steel. In the absence of high chloride levels the corrosion specimens exposed under bridges showed lower corrosion rates than those directly exposed to the environment. In sheltered marine environments, the corrosion rates of the weathering steel were high, and severe pitting developed. Severe pitting was not apparent with the special mild steel.

In a laboratory test, Sydberger and Vannerberg (1972) studied the effects of relative humidity and corrosion products on the adsorption of sulfur dioxide on metal surfaces. The metals used in this experiment were iron (0.03%C-0.03%Si-0.13%Mn-0.013%Cu), zinc (0.021Fe-0.006Cu), copper (composition not given), and aluminum (0.1%Si-0.1%Fe-0.02%Cu-0.002Mn-0.03%Ti). This study found a marked dependence of sulfur dioxide level and humidity on the corrosion rate of metal surfaces. The study also found that no corrosion initiation occurs on a clean, polished iron surface at humidities below eighty percent and 0.10 ppm sulfur dioxide. It was concluded that it is possible to relate some of the corrosion sensitivity of different metals in polluted atmospheres to the adsorbability of sulfur dioxide on the metal surfaces.

Briggs (1968) exposed twelve carbon steels, low alloy cast steels, wrought steels, and malleable iron, with the compositions listed in Table II-C, to marine and industrial atmospheres for periods of one-, three-, seven-, and twelve-years and recorded weight loss information. This study concluded that unmachined cast surfaces, with the casting "skin" still intact, have no significant effect on the corrosion resistance of cast steels when compared to machined surfaces, regardless of the atmospheric environments. The fastest corrosion rates occur in the marine atmospheres, with lower but similar corrosion rates occurring at the industrial atmospheres and marine atmosphere 250 meters from the ocean.

The corrosion of cast steel decreases as a function of time, because the products of corrosion (scaling and rust coating) build up and act as a protective coating to the cast steel surface. The corrosion rate of the most resistant cast steel (containing two percent nickel) is less than that of lesser corrosion resistant cast steels at any given exposure time. Cast steels containing nickel, copper, or chromium as alloying elements have corrosion resistances superior to cast carbon steel and those alloys containing manganese when exposed to atmospheric environments. Increasing the nickel and chromium contents of cast steel increases the corrosion resistance in all three types of steel in atmospheric environments. All cast steels have greater corrosion resistance than malleable iron in industrial atmosphere and are superior or equivalent to the wrought steels in this environment. The corrosion rate in the marine atmosphere depends primarily on the alloy content. The atmospheric corrosion performance of plain-carbon cast steel is superior to the AISI 1020 wrought steel but is slightly inferior to malleable iron. Figure II-3 shows corrosion rate data of high-manganese cast steels. From the figure it is apparent that manganese additions to around 1.4% yields improved corrosion resistance in all environments.

Kilcullen and McKenzie (1979) evaluated the results of work carried out by British Steel Corporation on the corrosion performance of Cor-Ten A and Cor-Ten B. Mild steel (C=0.21-0.23) was used as the reference material. It was noted that the rusts formed on the weathering steels in either soil or water were not protective. Corrosion rates were similar to those experienced with mild steel. The atmospheric exposure results show that the corrosion rates of mild steel are significantly higher than Cor-Ten B, which corroded faster than Cor-Ten A steel. The range of corrosion rates during the last three years of the test was 9 to 63 mm/year for mild steel, 5 to 33 mm/year for Cor-Ten A steel and 7 to 36 mm/year for Cor-Ten B steel. The results of exposure of Cor-Ten B under bridges show reductions in corrosion rate with time, although to a lesser degree than in boldly exposed steel specimens (see Table II-D and Table II-E).

A report by Mannweiler (1968) covering findings on the evaluation of ferrous corrosion specimens (see Table II-F) that had been exposed for seven years to the atmospheres of five different test sites, including rural, marine, and industrial atmospheres, made the following conclusions. The rate of atmospheric attack was found to diminish with time. Of all fifteen materials, the unalloyed steel (AISI 1020) was the most strongly attacked. The copper-alloyed steel incurred about 43 percent less corrosion loss. Except for two malleable irons, all other malleable irons, ductile irons, and the copper-chromium-nickel alloy fell within a common average weight loss band -- 5.5 to 11 (mean = 7.5)

g/dm^2. Copper additions of about 0.5 percent to malleable iron improved the resistance to atmospheric attack by an average of about eighteen percent. Copper in rolled steel provided even more significantly improved corrosion resistance. Machining the malleable irons tended to result in higher corrosion; machining the ductile irons had the reverse effect, presumably through removal of the deleterious cast surface.

II.H. Corrosion Resistance of High Silicon Cast Iron

Peacock and Cangi (1973) completed a study of the corrosion resistance of high silicon irons, containing 14.5% silicon, 0.85% carbon, and 0.65% manganese. Some high silicon cast irons also contain 4.5% chromium. The chromium-free, 14.5% silicon alloy exhibited excellent resistance to all concentrations of sulfuric acid to the boiling point, hot, strong nitric acid and its vapors, dilute hydrochloric acid up to moderate temperatures, all commonly encountered organic acids, many salts, mild to moderate alkaline solutions, and most other common corrosives. A partial list of environments which can be successfully handled by the high silicon cast irons can be found in Table II-G. The chromium bearing alloy retains all the corrosion resistance of the chromium-free, high silicon iron. In addition, the chromium-bearing alloy possesses superior resistance to chloride environments and solutions containing strongly oxidizing constituents such as hydrochloride, chlorine dioxide, hypochlorites, and all types of chlorinated organic or inorganic solutions.

The high silicon cast irons are rapidly attacked by hydrofluoric acid or solutions containing fluoride compounds, with corrosion rates in excess of 0.5 inches per year. Sulfurous acid, sulfur dioxide, sulfur trioxide, some sulfides, and sulfites rapidly corrode these alloys. Also, hot, strong alkalis or alternating acid-alkaline conditions will also yield high corrosion rates of hundred of mils per year (1 mil = 0.001 inch).

II.I. High Chromium Alloy Cast Irons.

The high chromium (20-35% Cr) alloy cast irons are virtually immune to atmospheric corrosion. To develop outstanding atmospheric corrosion resistance, sufficient chromium must be present to combine with all carbon and leave a minimum of 13% chromium in solid solution (McCaul and Goldspeil, 1982). These materials behave in a manner similar to that of common austenitic stainless steels when exposed to atmospheric corrosion conditions. Typical atmospheric corrosion rates for these materials are in the range of micrometers (μm) per year.

II.J. Review of Long-Term Atmospheric Corrosion Studies.

Several experimenters have generated atmospheric corrosion data for long-term exposures. In one experiment (Larrabee and Coburn, 1961), 270 steels with three variations of chromium content, five of copper, two of nickel, three of phosphorus, and three of silicon were tested, and data for average loss of thickness after 15.5 years exposure to three different environments were tabulated (see Table II-H). The environments to which the steel specimens were exposed consisted of an industrial site (Kearny, New Jersey), a semi-rural site (South Bend, Pennsylvania), and a marine site (the 250 m lot at Kure Beach, North Carolina). Data from the Kearny site and the South Bend site were used to generate Figures II-4 and II-5 to show the effects of alloy content on the corrosion of steel. The investigators found that the greatest change in corrosion with a relatively small change in composition is that caused by an increase in copper content from 0.01 to 0.04 percent (steels Nos. 1 and 2, Table II-H). Improved corrosion resistance was also realized by increasing copper content of the steels (Nos. 3 and 5); however, this rate of improvement is less marked than that produced by the 0.01 to 0.04 percent increase. The data shows that the improvement in corrosion resistance realized from the increase in copper content from 0.01 to 0.04 percent is greater than an increase in a similar amount of any other element investigated. Improvements in corrosion resistance were also made by relatively small additions of nickel, chromium, silicon, and phosphorus, singly, but the greatest improvements are obtained by the addition of specific combinations of these alloying elements. The effects of each element were found not to be additive; therefore, the availability of data from long-time exposure tests of many steels is necessary for a rapid estimation of the corrosion resistance of steels with particular combinations of alloying elements. The experimenters state that previous unpublished work showed that the improvement in atmospheric-corrosion resistance of a steel is nearly linear with increasing nickel content.

In a review of data accumulated on commercial products under atmospheric corrosion tests, it was concluded that the copper content is the controlling factor in corrosion rates of low metalloid open hearth irons and steels (Larrabee, 1946).

In a series of experiments (Copson; 1945, 1948, 1952, 1960), 76 irons and steels were exposed for 15.5 to 18.1 years to an industrial and two marine environments. The 76 steels are arranged in thirteen groups based on composition. Some of the materials were substantially free of alloying elements, and in others, the alloy content ranged up to 5 percent. The main alloying elements in this study were copper, phosphorus, nickel, and chromium, but some of the steels also contained molybdenum and higher that normal amounts of manganese and silicon. The material compositions can be found in Table II-I.

Table II-J shows the average weight loss for each group of materials identified in Table II-I at each time interval. Table II-K shows the average pit depth for each group of materials identified in Table II-I. Table II-L shows average corrosion rate for each group of materials identified in Table II-I at each time interval. The average corrosion rates were calculated from the weight loss data. During exposure some of the specimens developed small perforations or large holes and some were attacked during cleaning. Also some of the materials were not exposed to all three environments. This accounts for gaps in the data tables. Due to the gaps in Table II-L, the data from Table II-K were used to calculate an average pitting rate in microns per year. Figures II-6 through II-15 were generated using the average pitting rates for each group of specimens.

Figure II-6 presents the average pitting rate as a function of time for high-purity iron plus copper alloys exposed to the industrial environment in the Copson studies for times as long as 18.1 years. Figure II-7 shows the average pitting rate as a function of time for low-phosphorus plus copper alloys exposed to the industrial environment. Figure II-8 shows the average pitting rate as a function of time for high-phosphorus plus copper alloys exposed to the industrial environment in the Copson studies for times as long as 18.1 years.

Figure II-9 indicates the average pitting rate as a function of time for high-manganese and silicon plus copper steel alloys exposed to the industrial environment in the Copson studies for times as long as 18.1 years. Figure II-10 shows the average pitting rate as a function of time for copper plus chromium and silicon steel exposed to the industrial environment. Figure II-11 shows similar results for copper plus molybdenum steel. Figures II-12 and II-13 show the average pitting rate as a function of time for nickel steel and nickel plus chromium and molybdenum steel exposed to the industrial environment, respectively. Similarly, Figures II-14 and II-15 show the average pitting rate as a function of time for nickel copper steel and nickel copper plus chromium steel exposed to the industrial environment, respectively. The data in these figures indicate that all of these steels appear to reach a saturation pitting rate of approximately 50 μm/year after periods up to 18.1 years.

Copson found the better low alloy steels were more durable at the industrial site than at the marine locations. Pit depths were not found to rate the steels in the same order as weight losses. Alloying was found to have decreased both pit depths and the number of pits, but the beneficial effect of alloying was more marked in decreasing weight losses than pit depths. Chromium and copper additions to steel alloys were found to be more helpful in the industrial environment. The pits in the industrial environment were not significantly deeper at 9 years than at 5 years. The pitting rate of alloyed steels appeared to decrease with time.

The 16 year exposure data (Southwell, et al., 1976) for more than fifty ferrous and non-ferrous materials (eleven materials important to this review have compositions as listed in Table II-M) in the tropical environments of the Panama Canal Zone, showed that copper-bearing steels were no more resistant than mild carbon steel in either a coastal location and an inland location. Aston-process wrought iron was equal to carbon steel in a coastal atmosphere, but was slightly less resistant in the inland atmosphere. The coastal atmosphere was about 1.3 times as corrosive to wrought steels as the inland location, but for cast iron and steel, there was little difference.

Eight year exposure studies (Alexander, et al., 1961; Forgeson, et al., 1958; Southwell, et al., 1958; 1960; and Southwell and Alexander, 1970) found that the corrosion rate of structural steel (see Table II-M) in the tropical marine environment is 1.5 to 2.2 times that at a temperate marine location (Kure Beach, North Carolina). Inland atmospheric corrosion in the Panama Canal Zone caused approximately 1/3 greater corrosion rates than in a typical U.S. industrial atmosphere. No difference in type or extent of corrosion was evident between pickled and millscale surfaces on mild carbon steel (metal A, Table II-M) in either of the tropical atmospheres. Copper-bearing steel was less effective in retarding corrosion in the tropical atmospheres (metal M, Table II-M) than in temperate zones (metal L, Table II-M). The corrosion of copper steel in the two tropical atmospheres was only 13 percent and 14 percent less than that of the unalloyed mild steel. Wrought iron (metal R, Table II-M), compared to four proprietary low-alloy structural steels (metals C, D, E, and F, Table II-M) in the tropical atmospheric environments, was much more heavily damaged by corrosion than any of the four steels.

III. Mechanisms of Aqueous Corrosion

The corrosion of iron and steel in the atmosphere, immersed in natural waters, or buried in soils, is controlled by aqueous corrosion mechanics. "General corrosion" is defined as the uniform attack of a metallic surface. Rates of general corrosion are reported in either micrometers penetration per year (µm/yr) or milligrams per square decimeter per day (mdd). Pitting and crevice corrosion are two types of localized attack. Localized attack is reported in terms of depth of penetration (µm). For localized attack, the rate of corrosion is greater at some areas than others.

The corrosion reaction at the surface of iron or steels is discussed by Uhlig and Revie (1985) and is summarized as follows. Aqueous corrosion of iron is governed by electrochemical theory and is related to a network of short circuited galvanic cells on the metal surface. Metal ions go into solution at anodic areas in an amount chemically equivalent to the reaction at cathodic areas. At anodic areas the following reaction takes place:

$$Fe^{+} \blacklozenge Fe^{++} + e^{-} \qquad \text{(III-1)}$$

This reaction is rapid, and the corrosion of iron is usually controlled by the slower cathodic reaction which, in dearated water, is

$$H^{+} \blacklozenge 0.5H_2 - e^{-} \qquad \text{(III-2)}$$

The cathodic reaction can be accelerated in the presence of dissolved oxygen, a process called depolarization, according to

$$2H^{+} + 0.5O_2 \blacklozenge H_2O + 2e^{-} \qquad \text{(III-3)}$$

The dissolved oxygen reacts with hydrogen atoms on the metal surface. The reaction proceeds as rapidly as oxygen reaches the metal surface. Adding equations III-1 and III-3 and using the reaction $H_2O \blacklozenge H^{+} + OH^{-}$ gives

$$Fe + H_2O + 0.5\,O_2 \blacklozenge Fe(OH)_2 \qquad \text{(III-4)}$$

Hydrous ferrous oxide, $FeO \bullet H_2O$, or ferrous hydroxide, $Fe(OH)_2$, comprises the corrosion product layer. The pH of saturated $Fe(OH)_2$ is about 9.5, so the surface of corroding iron is slightly alkaline. At the outer surface of the corrosion product layer, access to dissolved oxygen converts ferrous hydroxide to ferric hydroxide or hydrous ferric oxide in accordance with

$$Fe(OH)_2 + 0.5\,H_2O + 0.5\,O_2 \blacklozenge Fe(OH)_3 \qquad \text{(III-5)}$$

The corrosion product layer passivates the metal surface in that it resists the large thermodynamic driving force for corrosion (Uhlig and Revie, 1985). Two theories have been proposed regarding the nature of the passive film. The first, the "oxide-film theory," states that the corrosion product layer comprises a diffusion barrier which separates the metal from its environment, slowing down the reaction rate. The second theory states that the metal surface is covered by a chemisorbed film (oxygen) which displaces the normally absorbed water molecules and slows down the rate of anodic dissolution involving the hydration of metal ions. The passivity of iron and its alloys is believed to be due to the oxide-film theory.

General corrosion takes place on the entire surface of the metal. The rate is governed by the diffusion of oxygen through the passive layer. Pitting selectively attacks areas of the metal where there is 1) a break in the passive film (mechanically or chemically induced), 2) an emerging dislocation or slip step caused by applied or residual tensile stresses, or 3) a compositional heterogeneity such as an inclusion, segregate, or precipitate (Tretheway and Chamberlain, 1988). Pitting is unpredictable, with variable rates that depend on the migration of corrodents into and out of the pit. Pits may be shallow, deep, elliptical, undercut, or subsurface. The presence of chloride ions leads to pitting by dissolving local areas of the passive film (Jones, 1992). In carbon steels, it is believed that sulfide inclusions are responsible for pit initiation (Shreir, 1963).

The pitting reaction in carbon steel with MnS inclusions in a NaCl solution is presented by Shreir (1963) and is summarized as follows. Within the pit, the primary anodic reaction is given by Fe^+ ♦ $Fe^{++} + e^-$ followed by $Fe^+ + H_2O$ ♦ $Fe(OH)^+ + H^+$, which reduces the pH within the pit. At the pit mouth, a membrane of magnetite (Fe_3O_4) and rust (FeOOH) is formed, that prevents the mixing of the acidic electrolyte within the pit with the more alkaline electrolyte outside the pit. The outside of the pit becomes passivated, and corrosion is concentrated within the more acidic pit.

Crevice corrosion is the localized attack of the metal surface within a crevice and is encountered when halide ions (Cl^-, F^-, or Br^-) are present within the crevice. The mechanism of crevice corrosion is presented by Shreir (1963). Initially the whole surface (including the crevice) undergoes uniform attack. The metal surface is readily accessible to free oxygen. The access of dissolved oxygen in the crevice is limited by diffusion. The oxygen concentration within the crevice rapidly becomes negligible and the cathodic reaction within the crevice ceases. The cathodic reaction on the free surface results in anodic dissolution of the metal within the crevice. The excess of positive ions generated in the crevice results in the migration of halide ions and OH^- into the crevice. The small concentration of OH^- does not affect the pH within the crevice. The liberation of H^+ ions reduces the pH to about 3 within the crevice. In combination with the presence of the halide ions, the liberation of H^+ ions prevents the development of the passive layer. This sequence of events results in an autocatalytic process.

III.A. Environmental Factors That Influence Corrosion in Natural Waters

The following is a brief review of the factors that influence corrosion in natural waters (Uhlig and Revie, 1985). The corrosion rate of irons and steels in natural waters depends directly upon the concentration of dissolved oxygen present. The corrosion rate increases as the concentration of dissolved oxygen increases until a critical point is reached, where the corrosion rate begins to decrease. The increase is caused by the increased concentration of oxygen that can diffuse through the passive layer. In fresh water, as the oxygen concentration increases further, more oxygen is present at the metal surface than can be reduced. The excess is available to further oxidize the corrosion product layer, leading to a more passive film. At high oxygen concentrations, breakdown of the passive layer can lead to severe pitting. Due to the presence of chlorides in seawater, passivity of iron does not occur. The corrosion increases continually with increasing oxygen concentrations. In distilled water, the critical oxygen concentration is about 12 ml/liter and increases with increasing dissolved salts and increasing temperature.

The corrosion rate of iron doubles for every 30°C increase in temperature. Temperature increases lead to more rapid corrosion reactions at the metal surface. In an open system, which allows oxygen to escape, the corrosion rate increases to a temperature of about 80°C, and then falls to a low value at the boiling point. The decrease at 80°C is attributed to a decrease in the oxygen solubility in water, which overwhelms the effect of the temperature increase.

The corrosion rate of iron is independent of pH in the range of 4-10. The passive layer is being replenished due to corrosion. In this pH range, the surface of the iron is always in contact with an alkaline solution of saturated hydrous ferrous hydroxide, with a pH of about 9.5. For pH less than 4, the diffusion layer dissolves and the metal surface is in direct contact with the environment. For pH above 10, the increase in alkalinity further increases the passivity .

Increasing the NaCl concentration will first lead to an increasing corrosion rate followed by a decreasing rate when the concentration exceeds about 3 wt%. The increase is attributed to a change in the protective nature of the diffusion barrier film. Water with NaCl present has a greater conductivity than pure water. As such, anodes and cathodes present on the metal surface can operate further from each other. At these cathodes, NaOH does not immediately react with $FeCl_2$ which diffuses from the metal surface into solution. $Fe(OH)_2$ is thus formed away from the metal surface. The decreasing corrosion rate observed with increasing NaCl concentration above 3 wt% is caused by the decrease in the solubility of oxygen in water.

Alkali metal salts (KCl, LiCl, Na_2SO_4, KI, NaBr) affect the corrosion rate of iron and steel in approximately the same manner as NaCl. Chloride compounds appear slightly more corrosive in the order LiCl, which is greater than NaCl, which is greater than KCl. Alkaline earth salts ($CaCl_2$, $SrCl_2$) are slightly less corrosive than alkali metal salts. Nitrides appear less corrosive than sulfides or chlorides at low concentrations. Acid salts ($AlCl_3$, $NiSO_4$, $MnCl_2$, $FeCl_2$) lead to increased corrosion rates combined with hydrogen evolution and oxygen depolarization. Alkaline salts (Na_3PO_4, $Na_2B_2O_7$, Na_2SiO_3, Na_2CO_3) act as corrosion inhibitors. Oxidizing salts are either corrosives ($FeCl_3$, $CuCl_2$, $HgCl_2$) or inhibitors (Na_2CrO_4, $NaNO_4$, $KMnO_4$, K_2FeO_4).

Natural fresh water contains dissolved calcium and magnesium salts. The presence of these compounds decreases the corrosion rate of iron and steels by the deposition of a thin diffusion barrier film (largely $CaCO_3$) on the metal surface. This film limits the diffusion of oxygen and supplements the natural $Fe(OH)_2$ passive layer. The ability to form this supplemental diffusion layer depends on the ability to precipitate $CaCO_3$ on the metal surface. This depends on the alkalinity or acidity of the water, the pH, and the concentration of dissolved solids in the water.

The presence of small amounts of carbon dioxide in hard waters affects the corrosion rate of iron and steels by determining whether the protective layer of $CaCO_3$ will precipitate. Waters containing CO_2 in excess of that required to keep calcium carbonate in solution are more corrosive than those containing just enough CO_2. Carbon dioxide in appreciable concentrations accelerates the corrosion rate of iron, as carbonic acid is similar to any acid that reacts with iron (Shreir, 1963).

In some natural waters the corrosion of iron has been observed to occur at a higher rate than expected. These high rates have been traced to the presence of sulfate-reducing bacteria. In the presence of hydrogen or organic matter, these bacteria readily reduce inorganic sulfates to sulfides. The presence of an iron surface aids the corrosion-process by supplying hydrogen, which the bacteria use to reduce (SO_4^{2-}) ions. The reaction sequence is outlined as follows (Uhlig and Revie, 1985).

Anode: $4Fe \blacklozenge 4Fe^{++} + 8e^-$

Cathode: $8H_2O \blacklozenge 8H_{ads.\ on\ Fe} + 8OH^- - 8e^-$

$8H_{ads} + Na_2SO_4 \blacklozenge 4H_2O + Na_2S$ (in presence of bacteria)

$Na_2S + 2H_2CO_3 \blacklozenge 2NaHCO_3 + H_2S$

Summary: $4Fe + 2H_2O + Na_2SO_4 + 2NaHCO_3 + 2H_2CO_3 \blacklozenge$

$3Fe(OH)_2 + FeS + 2NaHCO_3$

Severe damage by sulfate-reducing bacteria has been observed in oil well casing, buried pipelines, and in pipe from deep water wells.

III.B. Metallurgical Factors That Influence Corrosion in Natural Waters

The corrosion of iron or steel in natural waters is controlled by oxygen diffusion. During hot rolling, millscale (oxide) will occur on the metal surface. In water, severe pitting may occur if large amounts of millscale are present (Shreir, 1963).

All carbon steels, low alloy steels, and wrought irons corrode at the same rate when immersed in natural waters. Corrosion is reduced only when a steel is alloyed in the proportions of a stainless steel (Cr greater than 12%), a high-silicon iron, or high-nickel iron alloy. In these cases, oxygen diffusion is no longer the controlling mechanism (Uhlig and Revie, 1985). Other alloying elements (C, N, Mn, S, P, Cu, Ni) do not appear to influence the corrosion rate of immersed steels (Uhlig, 1948). Serious pitting may occur in the area of welds, and is attributed to galvanic effects between the weld metal and the steel plate. Use of noble electrodes for welding can overcome this problem (Shreir, 1963).

It appears that heat treatment is a minor factor with respect to corrosion of carbon steels in natural waters. There is evidence that carbon steel quenched from high temperatures (850°C) corrodes at a higher rate than steel quenched and subsequently tempered. A 0.38% carbon steel quenched from 850°C and immersed in quiet distilled water corroded at a higher rate (0.0043 ipy) than a tempered steel (0.0033 ipy) immersed in similar conditions (Uhlig, 1948).

The corrosion of iron is retarded in most aqueous mediums when it is coupled to zinc or magnesium. An exception for zinc may occur at elevated temperatures in fresh water where the zinc becomes less effective and even cathodic. In seawater, iron is protected when coupled with zinc, magnesium, cadmium, or aluminum. Corrosion of iron in seawater is accelerated when it is coupled with copper, chromium, nickel, and other metals of more noble galvanic potential (Uhlig, 1948).

The corrosion behavior of cast irons differs from that of steels. A steel will undergo uniform attack with its overall dimensions steadily reducing. Cursory examination of a cast iron may suggest that it has not corroded at all. This is due to the fact that cast irons contain in their microstructures several more or less corrosion resistant components that are absent from steels (namely graphite, phosphide eutectic, or carbide). If a cast iron corrodes in a manner where the corrosion product layer is deposited some distance from the surface, a skeleton comprising stiffened graphite flakes will be left behind. This process is known as "graphitization". Small variations in the composition of cast irons or the addition of small amounts of alloying agents has little effect on the corrosion resistance of cast irons (Shreir, 1963).

The addition of about 20% nickel to cast iron produces a material with a stable austenitic structure and is known as austenitic cast iron. Austenitic irons have better corrosion behavior in natural waters (0.02 - 0.058 mm/yr) than ferritic irons (0.066 - 0.53 mm/yr) (Shreir, 1963).

Cast iron can be made corrosion resistant by adding chromium, in the range of 25-35%. A useful empirical formula for determining the minimum chromium content for corrosion resistance is given by Shreir (1963) as

$$\% \text{ Cr} = (\% \text{ C} \times 10) + 12$$

These alloy cast irons owe their corrosion properties to the formation of an impervious and highly tenacious film consisting of a mixture of iron and chromium oxides. The chromium oxide is formed from chromium in the matrix and not from that combined with the carbide. As such, "stainless iron" will only be formed when enough chromium (about 12%) is dissolved in solid solution in the matrix. Increases in silicon or molybdenum contents may increase the corrosion resistance of high chromium iron. This type of iron is most useful in acid waters, for example mine waters, containing oxidizing agents that are corrosive to other cast alloys. Abrasion may damage the protective film, resulting in localized corrosion (Shreir, 1963).

The normal silicon content of cast iron (up to about 3%) has little effect on the corrosion resistance. An increase in the corrosion resistance occurs for silicon contents greater than 3%, and is attributed to the development of a corrosion resistant film with a large proportion of silica. The full protective value of the film does not develop until at least 14.25% of silicon is present. Increasing the silicon content above this value does not effect the corrosion resistance (Shreir, 1963). An alloy of 15%Si-Fe is generally anodic to nickel, copper, and carbon. This alloy is cathodic to zinc and aluminum. Alloying additions such as molybdenum, tungsten, vanadium, or columbium do not contribute significantly to the corrosion resistance of silicon-iron (Uhlig, 1948).

III.C. Aqueous Corrosion Studies

A review of the technical literature with respect to aqueous corrosion of carbon steels, cast irons, low alloy steels, and alloy cast irons has revealed several studies that might have direct application to materials degradation under repository-relevant conditions. These studies encompass a wide variety of exposure conditions including seawater (Southwell and Alexander, 1970; Larrabee, 1953; Hudson and Stanners, 1955; Peterson and Waldron, 1961; Hadfield and Main, 1935; Wheatfall, 1967; Reinhart, 1965; and Fink and Boyd, 1970), fresh water (Southwell and Alexander, 1970; Larrabee, 1953; and Coburn, 1978), and ground water (Marsh and Taylor, 1988; McCright and Weiss, 1985; and Hong and Pitt, 1987).

The most extensive aqueous corrosion study that included candidate materials was completed by Southwell and Alexander (1970). The study was undertaken to identify materials suitable for application in the harsh environment of the Panama Canal Zone. The experiment, which included atmospheric and aqueous exposure of numerous materials, lasted sixteen years. Materials tested of interest are carbon steels with various surface finishes, copper bearing steel, nickel bearing steel, chromium bearing steel, low alloy steels, cast steels, gray cast iron, wrought iron, and austenitic (high nickel) cast iron. The compositions of the materials tested are provided in Table III-A.

Aqueous corrosion testing consisted of exposure of metal samples to three different environments. Samples were exposed for periods of 1, 2, 4, 8, and 16 years to constant immersion in sea and lake water. Samples were also subjected to intermittent immersion in sea water by mounting the samples at the mean tide level. Redundant samples were employed to permit removal of the representative specimens at the time intervals listed above. The results of these tests are provided in Tables III-B, III-C, and III-D and are summarized below.

In all three environments, the corrosion rate decreased with increasing exposure time. The pitting rate also decreased with increasing exposure time. Figure III-1 shows the general corrosion rate as a function of exposure time for the carbon steel under sea water immersion, mean tide, and fresh water immersion, respectively. Figure III-2 shows the maximum pit depth as a function of exposure time for carbon steel under sea water immersion, mean tide, and fresh water immersion, respectively. Figures III-3, III-4, and III-5 show the maximum penetration rate for the materials with finishes as listed in Table III-A under fresh water immersion, sea water immersion, and mean tide immersion, respectively.

The average penetration rate of carbon steel with millscale on its surface was slightly less than the same steel with a machined or pickled surface. The presence of millscale lead to severe pitting when immersed in sea water, with complete penetration occurring in less than eight years. Carbon steel with a machined surface showed a greater susceptibility to pitting than that with a pickled surface.

Carbon steel with a pickled surface corroded at a greater rate when immersed in sea water (73µm/yr vs 45 µm/yr in fresh water). No difference was noted in the corrosion rate of the materials exposed to mean tide and complete immersion conditions. Descending order of pit depths for 16 year exposure were: sea immersion, 3937 µm; mean tide, 2489 µm; fresh water, 2362 µm.

The addition of up to 5% Ni to carbon steel did not result in increased general corrosion resistance, but led to more severe pitting under complete sea water immersion. The addition of up to 5% Cr to carbon steel increased the corrosion resistance only in fresh water. The pitting behavior of Cr-containing steels was similar to carbon steel in the three environments. The addition of Cr reduced the corrosion resistance of steel when immersed in sea water and lead to more severe pitting under mean tide conditions.

Low alloy steels did not demonstrate superior corrosion resistance over carbon steels for any of the three exposure conditions. For several alloy-steels, the average corrosion rate and the pitting behavior were worse than for carbon steels, as shown in Table III-A.

Wrought iron appeared to have better corrosion resistance than carbon steels with machined surfaces in all three environments (8 yr tests, fresh: 47.6 µm/yr vs 63.5 µm/yr; immersion: 69.9 µm/yr vs 85.7 µm/yr; mean tide: 50.8 µm/yr vs 79.4 µm/yr). The pitting behavior of wrought iron when compared to carbon steel is similar, except for mean tide immersion, where wrought iron demonstrated superior pitting behavior (1245 µm vs 3150 µm). The 8 year pitting value of machined carbon steel is questionable because it is inconsistent with the one and 16 year data.

Cast steels with a machined surface corroded at approximately the same rate as wrought carbon steel in all three environments. The pitting of cast steels was found to be worse than that of wrought carbon steel, however. Gray cast irons had the worst corrosion and pitting behavior of all the ferrous metals tested in all three environments. Austenitic cast iron demonstrated the best general corrosion and pitting performance (except under mean tide exposure) in all three environments, as is shown in Table III-E.

Another test of several steels immersed in sea water was conducted at Kure Beach, North Carolina (Larrabee, 1953). The data from this test again demonstrates that the corrosion rate decreased with increasing exposure time, ranging from 76 to 96 µm/yr for several materials, as shown in Table III-F. Figure III-6 shows the corrosion rate data of several weathering structural steel following immersion in sea water at Kure Beach. This figure shows an improvement in corrosion resistance with small alloying compositions.

The corrosion rate of structural carbon steel was less than that observed at the Panama Canal Zone (83.8 µm/yr - 4.5 years, Kure Beach vs 101.6 µm/yr - 4 years, Panama). This is attributed to the lower sea water temperature at Kure Beach.

Only a 2.6% Cr-1% Mo steel showed any beneficial corrosion resistance after exposure for 4.5 years at Kure Beach. The results observed at Panama indicate that for longer exposure periods Cr additions up to 5% yielded no improvement in the corrosion resistance. None of the other alloys tested at Kure Beach demonstrated improved corrosion resistance.

Sea water immersion tests that included candidate container materials or materials similar to candidate container materials were conducted by the British Iron & Steel Research Association at Emsworth, England for periods of up to five years (Hudson and Stanners, 1955). The purpose of these tests was to investigate the effects of small amounts of alloying elements on the corrosion resistance of mild steel. Specimens were immersed in a large tank of water which was situated on the shore. The tank filled with water at each tide and the water remained in the tank when the tide receded, giving stagnant conditions. The compositions of the materials tested and a summary of the results are provided in Table III-G. Figure III-7 shows that, keeping copper content constant, increases in manganese result in decreased long-term corrosion rates.

The heat treatment of the specimens had no effect on the corrosion resistance. The corrosion rate was found to remain constant or decrease with increasing exposure time. The carbon content was found to have no effect on the corrosion resistance. In contrast with the tests conducted at the Panama Canal Zone, the corrosion of steels bearing 5% Cr was found to be superior to mild steel. It must be noted that for the 16 year sea water immersion tests in Panama, the corrosion resistance of 3% Cr bearing steels was worse than that of mild steels (70 µm/yr vs. 33.5 µm/yr). At Emsworth, an alloy containing 1.4%Al-2.8%Zr-0.8%Si-0.06%Cu proved to be the most corrosion resistant alloy tested (24.9 µm/yr vs 61.5 µm/yr for carbon steel). The addition of other alloying elements showed no significant benefit. The general corrosion of immersed cast irons did not differ from that of plain carbon steel, but immersed cast irons suffered a greater degree of localized corrosion and graphitic corrosion.

The range of corrosion rates observed at Emsworth are less than those observed at either Kure Beach or the Panama Canal Zone. This is attributed to a lower water temperature.

Extensive sea water immersion and mean tide immersion tests of several alloys were conducted at various locations for a period of five years. These locations were Halifax, Nova Scotia; Auckland, New Zealand; Colombo, Ceylon; and Plymouth, England (Hadfield and Main, 1935). The compositions of the materials tested and a summary of the results are provided in Tables III-H and III-I. The addition of 36% Ni resulted in improved general corrosion resistance at all four locations. Pitting was found to be more severe under complete immersion at Halifax. For all other combinations of locations and environments, the pitting behavior was improved relative to the Halifax results. The addition of 13% Cr resulted in improved general corrosion behavior. Pitting was found to be more severe at Halifax (immersion) and Colombo, Ceylon (immersion and mean tide). A summary of the data is shown in Table III-J. Other alloys had similar corrosion performance to carbon steel.

The rolled irons (wrought, Swedish, and ingot) showed similar behavior to carbon steel in all four locations under both mean tide and immersion conditions. The pitting behavior of rolled irons was found to be typically better than that of carbon steels for all combinations of location and exposure conditions with the exception of immersed Swedish charcoal iron and ingot iron at Colombo, Ceylon. In these materials, pitting was found to be worse.

Cast irons demonstrated better general and pitting performance than carbon steels. This is diametric to that observed in the 16 year tests at Panama, where cast iron was found to exhibit poorer corrosion behavior than carbon steel.

The general corrosion of carbon steel with and without a millscale layer present was similar, as shown in Table III-I. Pitting was found to be more severe with a millscale layer present.

Peterson and Waldron (1961) studied the corrosion rate of mild steel in San Diego Harbor. San Diego Harbor was considered to be corrosive due to the heavy contamination of the shallow bay by sewerage. The corrosion rate after 1.5 years was found to be 177 µm/yr vs 150 µm/yr in sea water at Panama. The pitting rate was found to exceed 1524 µm/yr (complete penetration of a sample in a one year period). The maximum pitting of mild steel in Panama was 1575 µm in one year. The general corrosion of mild steel in polluted sea water is greater than that in unpolluted sea water.

The corrosion of metals in deep ocean environments at depths ranging from 0 to greater than 5500 feet has been investigated (Wheatfall, 1967; Reinhart, 1965; and Fink and Boyd, 1970). The compositions of the materials tested and a summary of the results are provided in Table III-K. AISI 1015 steel exhibited a more uniform attack in deep ocean exposures than when exposed in shallow waters. The corrosion rate of AISI 1015 steel was found to be less at ocean depths (77 µm/yr) than in shallow waters (133 µm/yr) (Wheatfall, 1967). The corrosion rates of alloy steels at depths up to 5600 meters were found to be similar regardless of the steel composition. This corrosion rate ranged from 13.5 to 17 µm/yr for a 2.9 year exposure. The corrosion rate of alloyed steels immersed in the mud at the sea floor did not differ appreciably from water immersion at the same depth, as shown in Table III-K. The corrosion behavior of AISI 4340 alloy steel was found to not differ from other alloyed steels when immersed for approximately 0.5 year at 2340 feet (Reinhart, 1965). A comparison of the corrosion rates of 1010 carbon steel, AISI 1015 carbon steel, and other alloyed steels immersed for 2.9 years at depths greater than 5000 feet showed no differences.

The corrosion of metals in various fresh water environments including unpolluted temperate (Coburn, 1978), unpolluted tropical (Southwell and Alexander, 1970), and polluted temperate (Larrabee, 1953) is the subject of numerous studies. The metals tested and the results are presented in Table III-L. The corrosion rate of plain carbon, copper, and alloyed steels is greater in unpolluted tropical fresh water than temperate water (carbon steels: 47 vs 152.4 µm/yr; Cu-containing steel: 37 vs 152 µm/yr; alloy steels: 32 vs 100-125 µm/yr). These differences are attributed to the lower water temperature in the temperate climate. The corrosion of metal in temperate climates showed a dependence on the water condition. Of the temperate environments tested (Monongahela, Allegheny, and Mississippi rivers), the Monongahela River gave the highest general corrosion rates, as shown in Table III-L. This is due to the contamination of the river by ferric sulfate from mine drainage. Pitting in the Monongahela River, however, was not as severe as in the other two rivers. Stagnant conditions in the Allegheny and Mississippi rivers gave rise to oxygen concentration cells which lead to pitting. In all environments, the corrosion behavior of alloyed steels is similar to that of plain carbon steels. A summary of the corrosion of alloyed and plain carbon steels is shown in Table III-M.

Figures III-8 and III-9 show the average pitting rate and the maximum pit depth, respectively, for alloys with compositions as shown in Table III-L, after immersion in the Mississippi River. The Tri-Ten alloy steel is the only steel that shows a decrease in the average pitting rate form four to eight years (see Figure III-8), but conversely, it has the highest maximum pit depth at eight years (see Figure III-9).

The final area investigated in this study was the corrosion of metals in ground water. The materials tested and the results are provided in Table III-N. The corrosion of carbon steel A-36 and an alloy steel (9%Cr-1%Mo), was investigated in Utah geothermal waters (Hong and Pitt, 1987). These geothermal waters are neutral (pH=7), but contain large concentrations of dissolved solids, including 610 mg/l Na^+, 916 mg/l SO_4^{--}, and 460 mg/l HCO_3^-, and are considered aggressive. Corrosion rates of both A-36 and 9%Cr-1%Mo increased with increasing temperature to a maximum value and then decreased, as shown in Figure III-10. At low temperatures, the 9%Cr-1%Mo alloy steel corroded at a higher rate than the A-36. At higher temperatures the opposite was observed. Both materials suffered pitting with the A-36 steel having more severe pitting behavior than the 9%Cr-1%Mo steel, as shown in Table III-N.

The corrosion of forged and cast mild steels has been investigated in granitic ground water for an exposure time of 2.5 years (Marsh and Taylor, 1988). Samples were placed in tanks of simulated granitic ground water and covered with either Bentonite or crushed granite. Packing the samples in Bentonite reduced the corrosion rate by a factor of approximately three, as shown in Figure III-11. The reduction in the corrosion rate is due to the low oxygen permeability of the Bentonite. The thickness of the cover (ranging from 5-10 cm) was found to have no effect on the corrosion rate. The corrosion behavior of cast and forged steels was found be similar when covered with crushed granite (approximately 25 μm/yr), as shown in Table III-O. When covered with Bentonite, forge steel corroded at approximately 7 μm/yr while cast steel corroded at approximately 3 μm/yr.

Short duration (62.5 day) exposures of 1020 carbon steel, gray cast irons, 2.25%Cr-1%Mo alloy steel, and 9%Cr-1%Mo alloy steel were conducted in a simulated tuff ground water environment at varying temperatures (McCright and Weiss, 1985). The samples were covered with crushed tuff and air sparged to saturate the water with oxygen. All materials showed an increasing corrosion rate with increasing temperature, through a maximum between 80 and 90° C. Above these temperatures the corrosion rate decreased. Plain carbon steel showed the highest general corrosion rate and 9%Cr-1%Mo alloy steel showed the lowest (1020 C-steel at 90°C: 414 μm/yr vs 8.3 μm/yr for the 9%Cr-1%Mo alloy steel). Pitting was also less severe for the 9%Cr-1%Mo alloy, as shown in Table III-P. Figures III-12, III-13, and III-14 show the general, pitting, and crevice corrosion rates, respectively, for candidate corrosion allowance materials in J-13 water plotted with respect to temperature. In analyzing these figures, it is evident that the 2-1/4%Cr-1%Mo alloy is more resistant to general corrosion than plain carbon steel but suffers severe localized corrosion.

Potentiostatic anodic polarization studies have been conducted in a J-13 simulated nuclear waste repository environment, where the environment was allowed to evaporate to dryness followed by the rehydration prior to polarization (Wilde and Wilde, 1993). A summary of the conclusions drawn by the investigators follow. The anodic dissolution characteristics of Type 316L stainless steel in environments containing 1000 x Cl^- J-13 depend markedly on whether the solution is merely a mixture of virgin chemicals or a mixture that has been evaporated to dryness, baked and rehydrated to the same volume. In the former environment Type 316L stainless steel pits severely, whereas in the latter, a stable passive type of behavior is observed along with the precipitation of a dense white scale. Polarization of carbon steel in evaporated/rehydrated solutions resulted in qualitatively similar shaped curves and the deposition of a scale. X-Ray diffraction studies indicate that the scale on all specimens was calcium carbonate (Wilde and Wilde, 1993). Based on the observed qualitatively similar anodic dissolution characteristics of materials having widely differing corrosion behavior, it was concluded by the investigators that the factor controlling the polarization is the formation of a protective scale of calcium carbonate. It is the opinion of these investigators that future studies in Tuff type environments should be focused on solutions produced by an evaporation/rehydration cycle to better simulate what is anticipated in practice.

Results of many immersion tests in sea water show that the average penetration (calculated from weight-losses) of most plain and low alloy steels and irons is about100 μm/yr. Chromium steels in fresh water and in sea water exhibited less overall corrosion than plain carbon steel, but the tendency to pit is increased unless precautions are taken to prevent precipitation of chromium carbides. Corrosion can be very severe in the inter-tidal zone, particularly above high tide levels, where steels are wet periodically by salt water. Steels having low amounts of alloying elements deteriorate very rapidly under these conditions. While the corrosion rates of copper-containing and low alloy steels are considerably less under the same exposure conditions (Table III-I), these rates may still be too high for a satisfactory service life in a marine environment.

III.D. Summary of Aqueous Corrosion Results

Long term immersion tests in sea water and fresh water demonstrate that corrosion rates of low alloy steels are similar to plain carbon steels. Cast irons corrode at higher rates than plain carbon steels. The addition of less than 12% Cr or 36% Ni to carbon steel resulted in no beneficial improvements in corrosion resistance. In five year tests, steels with 12% Cr and 36% Ni demonstrated better general corrosion resistance, but in some cases pitting was more severe. High nickel "austenitic" cast irons were found to have the best corrosion resistance.

The packing of material in Bentonite was found to reduce the corrosion rate of mild steels by a factor of approximately three in a granitic ground water environment. Short term testing in a tuff ground water environment indicated that an alloy of 9%Cr-1%Mo provided improved corrosion resistance. An alloy of 5%Cr-0.5%Mo had improved general corrosion over plain carbon steel in fresh water (16 yr - Panama). In sea water, however, the general corrosion rate and pitting behavior were worse for the 5%Cr-0.5%Mo steel than the carbon steel.

Low alloy steels, plain carbon steels, cast steels, and cast irons corrode at similar rates. Alloys containing more than 12% Cr or more than 36% Ni corrode at a lower rate than plain carbon steels, but pitting may be worse. Short term tests indicate that a 9%Cr-1%Mo alloy steel may result in increased corrosion resistance. However, long term data could not be found. Austenitic cast irons show the best corrosion resistance under the conditions cited above.

A ranking of the aqueous corrosion performance of the materials investigated is provided below.

Austenitic Cast Iron > 12% Cr = 36% Ni = 9%Cr-1%Mo > Carbon Steel = Low Alloy Steels > Cast Iron

IV. Factors That Influence Corrosion in Soil

Soil presents a complex corrosion environment. Corrosion in soil is extremely variable with rates ranging from rapid to negligible as noted in Table IV-1. Corrosion in soil is governed by aqueous processes, which occur at the surface of the corroding material. The controlling mechanism is electrochemical. Corrosion rate in soils is controlled by the diffusion of dissolved oxygen, which is entrapped in the soil, through the corrosion product film on the surface of the metal. Soils having high moisture content, high concentrations of dissolved salts, and high acidity (low pH) are most corrosive. However, soil composition alone has little effect on corrosion resistance. The residence time of moisture on the metal surface controls the degree of corrosion, so soils with good drainage exhibit relatively low corrosion rates (Jones, 1992).

The corrosion rate in soils is also governed by factors that affect the access of oxygen and moisture. Aeration influences localized pitting corrosion. Aeration influences the solubility of the corrosion products, which dictates whether they provide a protective diffusion layer on the metal surface or diffuse into the soil. Local packing differences in a soil may also lead to the development of oxygen concentration gradients, which can lead to pitting (Romanoff, 1957).

IV.A. Soil Corrosion Studies

The National Bureau of Standards (NBS) has completed extensive studies of soil corrosion (Romanoff, 1957). The results of this study are summarized in this section. Table IV-A provides a limited sample of the results of the NBS study. This Table shows the corrosion of cast and wrought ferrous materials in select soils with differing aeration. Details of the NBS study can be obtained from Romanoff (1957).

Differences in the composition of plain carbon steels and unalloyed cast irons did not impact the corrosion rate in soils. Pitting of plain carbon steels and unalloyed cast irons is more severe in poorly aerated soils. In well-aerated soils, an ample supply of oxygen exists to promote the oxidation and precipitation of iron as ferric hydroxide near the metal surface. In poor aerated soils, the iron diffuses away from the metal surface into the soil matrix and no protective barrier is produced.

Field tests of low alloy steels showed that Cu-Mo open hearth irons, Cu-Ni low alloy-steels, Cr-steels, and Cr-Mo low alloy-steels exhibited better corrosion resistance than plain carbon steel. Pitting rates for the alloyed materials were at first more rapid than for plain carbon steels, but as time progressed became less than those of plain carbon steels. The relative corrodability of low-alloy irons and steels was not influenced by the soil aeration, with the exception of 4-6% Cr steels. In poorly aerated soils, these steels exhibited a greater pitting rate.

The NBS study (Romanoff, 1957) concluded that the corrosion of plain cast irons was not affected by the material composition but was affected by the soil conditions. The general corrosion rates of low alloyed cast irons are not significantly different than those of plain cast irons. For highly-alloyed cast irons (15%Ni-6.6%Cu-2.6%Cr), both general corrosion and pitting were substantially reduced relative to other materials.

The general corrosion of high-silicon (~14%) containing cast irons was found to be negligible in all but one soil (a clay). Comparisons of cast and wrought materials demonstrated that no difference in the general corrosion rate existed in a given soil. The pitting is more severe for cast materials in poorly aerated soils. In all soils, there were no differences in the pitting resistance. Removal of the millscale from hot-rolled materials resulted in slightly better corrosion performance.

V. Galvanic Corrosion

This section provides a brief review of galvanic corrosion issues related to carbon steels in environments relevant to the Yucca Mountain site. A more detailed evaluation of this topic will be the subject of a subsequent degradation modes survey.

A galvanic cell is formed when two dissimilar metals are connected electrically while both are immersed in a solution electrolyte. The fundamentals of galvanic corrosion of metals are discussed in Jones (1992), Shreir (1963), Wesley (1940), and LaQue (1951), and are summarized in this section. A metal or alloy has a unique corrosion potential, E_{corr}, when immersed in an electrolyte. When two dissimilar metals are coupled, the one with the more negative E_{corr} (active) has an excess of electrons. These electrons are lost to the more positive (noble) metal. In the couple, the anodic corrosion reaction is given as

$$M \blacklozenge M^+ + ne^- \qquad (V\text{-}1)$$

whose rate is increased by the loss of electrons. The more positive (noble) metal has the corrosion reaction

$$N \blacklozenge N^+ + me^-$$

(V-2)

which is decreased due to the presence of electrons drawn from metal M.

Figure V-1 shows the galvanic series for several alloys in sea water. The most active alloy (negative) in the couple is attacked preferentially by galvanic corrosion. A couple consisting of alloys with small potential difference will suffer galvanic corrosion to a lesser degree than a couple having a large potential difference. The galvanic series gives only tendencies for galvanic corrosion and provides no information regarding rates of attack. The galvanic series for sea water does not necessarily predict the behavior of galvanic couples in different environments, such as might exist in the proposed Yucca Mountain repository.

In aqueous solutions, the primary cathodic reaction is the reduction of dissolved oxygen at the metal surface according to

$$O_2 + 4H^+ + 4e^- \blacklozenge 2H_2O \qquad (V\text{-}3)$$

The maximum rate at which dissolved oxygen can be reduced is limited by the rate at which oxygen can diffuse to the cathode surface. Under stagnant conditions, the galvanic corrosion rate is under cathodic diffusion control. If the solution is not stagnant, the availability of oxygen is increased, resulting in an increased corrosion rate.

Galvanic corrosion is affected by the surface area of the anode relative to the cathode. A large ratio of cathode/anode surface areas will lead to enhanced corrosion of the anode. Galvanic corrosion is reduced when the cathode/anode ratio is small.

The rate at which oxygen can be reduced at the cathode also depends on the nature of the film adhering to the surface of the metal or so-called “passive layer.” If the cathode surface layer is not present or is unstable in a given environment, the reduction of oxygen occurs and the corrosion rate increases. Galvanic corrosion is reduced when the surface layer on the cathode is stable. Galvanic corrosion studies in flowing sea water (Shreir, 1963) revealed that steel corrodes more readily when coupled to either copper or nickel as compared to couples with either titanium or stainless steel. This is due to the presence of a very stable passive layer on the cathodes of titanium or stainless steel.

The composition of the electrolyte affects the distribution of galvanic corrosion on the anodic surface. When the electrolyte is of relatively high electrical conductivity, such as sea water, the galvanic corrosion on the less noble anode is distributed uniformly, and the weight loss is relatively high. As the electrical conductivity of the electrolyte is reduced as in fresh water, galvanic corrosion is concentrated near the bimetallic junction, and weight loss is reduced. The effects of cathode/anode area differences are not significant when the electrolyte has a very low electrical conductivity.

Under atmospheric corrosion conditions, the electrolyte is a thin film of moisture containing soluble contaminants (pollution, chlorides). Atmospheric galvanic corrosion typically occurs within a region 12-25 millimeters away from the bimetallic junction. The effect of differing anode and cathode areas do not effect atmospheric galvanic corrosion. Atmospheric galvanic corrosion is highly variable and depends heavily on changing atmospheric conditions (periodic wetting/drying, solute and particulate concentrations). The presence of sulfates in the atmosphere tend to accelerate atmospheric galvanic corrosion.

As a part of the sixteen year corrosion study conducted at the Panama Canal Zone discussed earlier, galvanic couples of several materials being considered as candidate materials were exposed to various environments (Southwell, et al., 1976). These environments were sea water immersion, mean tide, fresh water immersion, coastal atmosphere, and inland atmosphere.

For immersion testing, strips (2 x 9 x 0.25 inches) were coupled to plates (9 x 9 x 0.25 inches). Samples were removed and analyzed after exposures of 1 year, 8 years and 16 years. The results are provided in Table V-A and are summarized as follows.

For strips coupled with carbon steel plates (0.24% C), reduced corrosion of the strips was observed in all three aqueous environments except for couples with plates of 99% Cu, 99% Al, wrought iron, and zinc. Carbon steel plates corroded less in sea water when coupled with nickel alloy steel (slightly), 99% Al, wrought iron, and zinc. Under mean tide conditions, carbon steel plates were found to corrode slightly less when coupled with strips of Type 316 stainless steel, nickel, 70Ni-30Cu, low alloy brass, naval brass, 99% Cu, 99% Pb, and wrought iron. Couples of carbon steel strips and plates of Al and Zn were found to reduce the corrosion of the strips. Carbon steel plates were observed to corrode more in fresh water except when coupled with Zn and 99% Al.

Carbon steel strips corrode less in all three aqueous environments except when coupled with plates of 99% Al, galvanized steel, wrought iron, and 0.3% Cu steel. The corrosion of the following plates was observed to be worse when coupled with carbon steel plates in certain environments as listed in Table V-B.

Zinc strips coupled to various ferritic metal plates lead to enhanced corrosion of the strips and reduced corrosion of the plates in all three aqueous environments.

Phosphor Bronze (4%Sn-0.25%P) strips coupled with ferritic metal plates resulted in a reduced corrosion of the strips and enhanced corrosion of the plates in all three aqueous environments. The only couple showing different behavior was a couple with 70%Ni-30%Cu.

For atmospheric galvanic corrosion testing, discs of approximately equal area were coupled together. Corrosion of these couples was only around the edges. Corrosion in these exposures is not related to average penetration, but weight loss. A summary of the findings of atmospheric corrosion testing follows.

Stainless steels, bronzes, gray cast iron, Cu-containing steel, and a Cr-Ni-Mn low alloy steel were found to be noble with respect to carbon steel. Carbon steel was found to be slightly protected when coupled with nickel-silver, most brasses, lead, 5% Ni-steel, Cu-Ni-Mn-Mo low alloy steel, wrought iron, magnesium, aluminum and its alloys, zinc, and galvanized steel. Phosphor bronze was found to be protected by carbon steel, cast steel, and low alloy steels. Cast irons, cast steel, and Ni-steel were found to be protected by zinc.

In summary, for aqueous galvanic couples, carbon steel is found to be more active than the majority of the metals it is coupled with, exceptions being aluminum alloys and pure copper. Carbon steel was observed to be more active when coupled with stainless steels, low alloy steels, brasses, and bronzes (except Al-bronze). Zinc was found to be more active coupled to carbon steel, low alloy steels, cast steel, and austenitic cast iron. Phosphor bronze is noble with respect to carbon steel, low alloy steels and cast irons (except 70Ni-30Cu). No information was presented regarding the coupling of low alloy steel and irons with each other or non-ferritic materials.

Recent studies of galvanic corrosion were made at the Panama Canal Zone (Pelensky, et al., 1976; 1978). In these studies, various low-alloy steels were coupled to several metals. The results of these studies are shown below (see Tables V-C and V-D), and are summarized as follows. AISI 4340 experienced accelerated corrosion when coupled to stainless steel and Monel in the atmosphere. In soil exposures, a slight increase in the corrosion of ASTM 4340 steel was observed when coupled with brass, Monel, and 6061 aluminum alloy. For sea water immersion, an increase in the AISI 4340 corrosion rate was observed in couples with stainless steel, brass, Monel, and a titanium alloy. Etching of the steel showed a pocked surface in all three environments.

Twenty-four hour laboratory tests of the coupling of aluminum alloys (1100, 2024, 2219, 6061, 7075) with AISI 4130 low-alloy steel demonstrated marginal compatibility (Mansfield and Kenkel, 1976). The conditions tested were 3.5% NaCl in distilled water, distilled water, and tap water.

Tests in sea water demonstrated that increasing the cathode/anode area of Cu in a Cu-carbon steel couples led to enhanced protection of the carbon steel member. Small cathode/anode area ratios (Cu:Steel up to 1:1) led to no protection of the carbon steel (Compton, 1976).

Studies of stressed stainless steels in soils demonstrated that couples with iron results in failure (cracking) of only Type 301 stainless steel. Other stainless steels (Type 304, 26%Cr-1%Mo, and 26%Cr-6.5%Mo) did not show increased corrosion. It should be noted that Type 301 stainless steel also failed in several soils when coupled with magnesium and zinc. The failure is probably due to hydrogen embrittlement (Escalante and Gerhold, 1976).

VI. Stress Corrosion Cracking

Stress-corrosion cracking (SCC) is proposed as a process in brittle fracture whereby the surface energy associated with crack formation is reduced by the adsorption of atom or ion species in the stress-corrosion medium (Coleman, et al., 1961). Embrittlement by stress corrosion is most commonly observed in the form of time-dependent brittle fracture when a metal or alloy is subjected to static load while in a particular liquid or gaseous medium (Coleman, et al., 1961). The dominant mechanism for SCC is thought to be hydrogen embrittlement.

VI.A. Hydrogen Embrittlement

In a study of hydrogen embrittlement of mild steels in hydrogen sulfide solutions, Kawashima, Hashimoto, and Shimodaira (1976) found that the susceptibility to hydrogen embrittlement in hydrogen sulfide solutions is higher, the lower the strain rate and pH, and the higher the cathodic polarization. The change in concentration of hydrogen sulfide has little effect on the susceptibility to hydrogen embrittlement. They found that the susceptibility to hydrogen embrittlement is highest at 50°C and that the susceptibility to cracking increased with the increase of hydrogen absorbed in steel. Consequently, the absorbed hydrogen contributes to hydrogen embrittlement of mild steel as the sum of the applied stress and the apparent stress from the absorbed hydrogen. The failure will occur even at very low applied stress if the amount of absorbed hydrogen increases considerably (Kawashima, et al., 1976). This study estimated the activation energy for hydrogen embrittlement of mild steel in corrosive environments to be 8800 cal/mol. This study also concluded that the dissolved species that promotes hydrogen embrittlement in hydrogen sulfide solutions is undissociated molecular H_2S. The undissociated molecular H_2S, adsorbed on steel, acts as a bridge-forming ligand for the proton discharge, which accelerates the discharge reaction and hydrogen entry into steel, promoting hydrogen embrittlement.

Bruckner and Miles (1959) studied the susceptibility of cathodically protected steel to hydrogen embrittlement. They found that the presence of a reducible material such as FeO in the mill scale delays the uptake of hydrogen by the base metal during cathodic protection in the absence of sulfide ion contamination. In the presence of sulfide ions and low pH, they found that reduction of FeO in the mill scale and hydrogen uptake of the base metal may occur simultaneously during cathodic protection. They found that the removal of FeO from the base metal surface by reduction, due to reaction with hydrogen during cathodic protection may produce embrittlement in bend tests of certain steels as a consequence of geometric or notch effects. The susceptible steels are those in which the mill scale has penetrated grain boundaries during hot rolling, or the mill scale was rolled into the surface.

Whenever mild steel is stressed in tension to levels near or beyond the elastic limit, and exposed simultaneously to hot concentrated alkaline or hot concentrated nitrate solutions, the steel develops intergranular cracks. This phenomenon is called intergranular stress-corrosion cracking (IGSCC). For IGSCC to occur, it is necessary to have a tensile stress acting together with a specific environment. Bernstein and Thompson (1976) state that the evidence for hydrogen embrittlement as the dominant mechanism for stress corrosion cracking in mild steel alloys is now so strong as to place a rather heavy burden of proof on any proposed alternative.

In examining hydrogen cracking in mild steel, Pumphrey and Akhurst (1977) concluded that the effect of Fe_3C on hydrogen cracking in mild steel depends on the form of the carbide. In furnace cooled material, pearlite colonies are the major source of cracks. However, in quenched and tempered material, cracks are nucleated at carbides only as a result of the proximity of primary cracks nucleated at inclusions. This process accelerates the overall rate of cracking. The magnitude of this effect of tempering is the net result of these two opposing influences.

Payer, Preban, and Leckie (1976) found that susceptibility of low carbon martensitic steel to hydrogen stress cracking, as indicated by shorter times-to-failures at a given stress level and lower threshold stresses, increased with increasing ultimate tensile strength over a range 160 to 246 ksi (1100 to 1700 MN/m^2). Further, they found that the threshold stress, below which hydrogen stress cracking does not occurs, decreases with time and that cold working or tempering affects susceptibility in a manner not determined by this experiment.

Banerji, Briant, and McMahon (1977) showed that susceptibility of low alloy steels to hydrogen-induced intergranular cracking at low stresses is directly dependent on the amount of grain boundary impurity segregation. With increasing strength, the impurity effects become extremely severe, and smaller amounts are sufficient to induce embrittlement. In the absence of such segregation, transgranular hydrogen cracking occurs at very high stresses. Thus, they conclude that hydrogen-induced cracking in such steels would be transgranular and occur at high stresses were it not for the segregation of residual impurities to the grain boundaries which causes intergranular weakness.

VI.B. Sulfides

Schuetz and Robertson (1957) presented a summary of the experimental data developed by and conclusions drawn from a research project on sulfide stress-corrosion cracking sponsored by the National Association of Corrosion Engineers at Yale University. The materials used in the experiment were a set of relatively pure iron-nickel alloys varying from zero to thirty percent nickel. The carbon content was also a consideration in iron-carbon alloys. This study concluded that the basic cause of failure in sulfide environments is associated with the absorption of hydrogen leading to hydrogen-embrittlement. This is in agreement with an experiment performed by Snape (1967), who studied sulfide stress-corrosion in medium and low-alloy steels. McGlasson and Greathouse (1959) concluded that all types of steels will crack in aqueous systems containing H_2S and that time to cracking failure is inversely related to the concentration of H_2S in solution. The principle factors determining failure in service are internal stress, the strength of the material, the magnitude of applied stress, and the hydrogen content (Snape, 1967 and Schuetz and Robertson, 1957).

Schuetz and Robertson found that both martensitic and ferritic microstructures fail when the stress level and the hydrogen content are sufficiently high. Cold working also increases the susceptibility of ferrite to failure. Failure was not produced in austenite by Schuetz and Robertson (1957). Snape (1967) found austenite-containing steels to be considerably more resistant than steels with body-centered cubic (BCC) ferritic structures. Hydrogen absorption from hydrogen sulfide does not depend on the microstructure of the substrate, provided composition remains constant. Hydrogen permeation rate is dependent on structure. It is a minimum in austenite and annealed Armco iron (completely ferritic). Data for austenitic microstructures indicated that hydrogen permeability decreased in time to small values. The hydrogen permeability of the martensitic microstructure of the same composition is high, and similar to that of cold rolled iron or 10 percent nickel martensite. Since the ferritic BCC microstructure is more "open" than the close-packed FCC austenitic microstructure, hydrogen is more mobile in ferrite than in austenite if the lattice structure contains no defects. However, the mobility of hydrogen in steels is also dependent upon imperfections in the lattice structure. A highly cold-worked steel or an alloy with a large fraction of impurities and lattice defects will exhibit a significant reduction in hydrogen permeability compared to the same alloy with a perfect lattice structure.

In analyzing the effects of alloy composition on sulfide stress-corrosion, Schuetz and Robertson (1957) found a substantial increase in permeability to hydrogen as the most pronounced effect of adding nickel to iron. For both hydrogen sulfide exposure and cathodic charging, the greatest hydrogen absorption is observed in Armco iron. Alloying with carbon or nickel decreases the saturation quantity with the exception of 30 percent nickel martensite which absorbs almost as much hydrogen as Armco iron when cathodically charged. Snape (1967) found no apparent influence on sulfide cracking by changes in alloy content with the exception of changes in carbon. Cracking susceptibility increased with increasing carbon content in the range of 0.13%C to 0.47%C. Snape attributed effects of alloying additions to the altering of the microstructure of the material. This study stated that an increase in elements which lower the martensite stabilization temperature M_s and stabilized the face-centered structure significantly improved the performance of the material relative to sulfide cracking.

In a different experiment performed by Snape, Schaller, and Forbes Jones (1969), the presence of untempered martensite was found to drastically reduce sulfide cracking resistance regardless of composition. The presence of untempered martensite in the microstructure was associated with a lack of resistance to sulfide cracking, irrespective of strength levels in the steels. These investigators found that, by an intercritical hardening treatment designed to introduce untempered martensite, and then tempering to eliminate this martensite, it was possible to simultaneously improve strength and sulfide cracking resistance significantly. This is due to the stabilization of an FCC microstructure in the tempering process which results in a higher hydrogen mobility and a greater resistance to sulfide cracking. The resistance to sulfide cracking displayed by steels with increasing nickel content was attributed to nickel lowering the temperature at which intercritical hardening was possible.

In analyzing the stress dependence of sulfide stress-corrosion, Schuetz and Robertson (1957) conclude that under constant stress and hydrogen concentration, fracture of ferrite or martensite is dependent on stress, hydrogen concentration, and time. Above an apparent limiting "endurance stress", fracture is time-dependent in the manner of the "static fatigue" of glass and other brittle solids. Schuetz and Robertson found that endurance stress is inversely proportional to the hydrogen concentration in the range of concentration investigated. The idea that cracking occurs when a minimum stress is exceeded is supported in a report by Parkins (1952). McGlasson and Greathouse (1959) found that there are hardness and applied stress minimums below which a given steel will not crack, and they found these minimums to be inversely related.

Snape (1967) found that resistance to cracking generally diminishes with increasing yield strength. Schuetz and Robertson (1957) found that cold working increases the saturation concentration of hydrogen and the internal stress with the result that failure is promoted in a constant deformation test. Snape concluded that normalized or normalized and tempered steels are less resistant than quenched and tempered steels. This conclusion is supported by Dvoracek (1970), who studied sulfide stress-corrosion on 24 steels with differing heat treatment and composition. Also, McGlasson and Greathouse (1959) found low alloy steels (0.46%C, 0.020-0.022%P, 0.036-0.70%S, 0.70-1.0%Mn, 0.10-0.20%Mo), quenched and tempered steels to be less susceptible to cracking than alloy steels for given stress levels. Dvoracek also noted that surface notches or flaws of depths greater than 0.025 inch with a length of 0.25 inch on acceptable steels can promote cracking.

Schuetz and Robertson (1957) found that a trace of sulfide ion in aqueous solution results in a 15-fold increase in the saturation limit of hydrogen absorbed by a nickel steel. The cause of catastrophic failures in hydrogen sulfide are associated with the promotion of hydrogen absorption characteristic of the sulfide ion. Similarly, increasing the hydrogen ion concentration of the environment increases hydrogen absorbed by the steel and promotes failure. Dvoracek (1970) found chloride to have no effect on sulfide cracking, but pH, hydrogen sulfide concentration, temperature, and cold working had pronounced effects. For example, a high-strength API Grade P-110 steel was acceptable, if it was quenched and tempered, provided that the pH was 8 or greater or the hydrogen sulfide concentration was below 75 ppm. Increasing the temperature in Dvoracek's experiment from 24°C to 194°C doubled the critical stress required for stress-corrosion cracking. in addition, this study found that cold working reduced the critical stress for stress corrosion cracking by 40%.

VI.C. Carbonates

Kowaka and Nagata (1976) exposed stressed specimens of mild and low alloy steels to CO/CO_2 environments to investigate transgranular stress-corrosion cracking. This study found that:

1. No stress-corrosion cracking of mild and low alloy steels occurred in water containing only CO or CO_2.
2. Transgranular stress-corrosion cracking occurred only in CO/CO_2 environments containing water.
3. Stress-corrosion cracking occurred over a wide range of CO and CO_2 ratios dissolved in water.
4. Susceptibility to stress-corrosion cracking of mild steel decreased with increasing temperature.
5. The cracking appears to be stress-corrosion cracking rather than hydrogen embrittlement.
6. The materials should be stress relieved by annealing.

Hixson and Uhlig (1976) examined the stress-corrosion cracking of mild steel in a carbonate solution and found that cold-rolling of the mild steel at and above 30% reduction of area provided resistance to SCC at -0.55 V (SCE). At the same potential, they found pearlitic steels containing 0.25% C or more to be resistant. With respect to alloy content, they found that low-carbon nickel-iron alloys up to at least 11.9% Ni are susceptible as are Si-Fe alloys up to at least 0.73% Si. Low alloy steels containing Cr and/or Mo in relatively low amounts (0.51%Cr to 3.04%Cr, 0.093%Mo to 5.70%Mo), exhibited reduced susceptibility to IGSCC. The crack-sensitive paths operating in carbonate environments appear to be the same as those operating in nitrate environment; i.e. concentration of carbon atoms at grain boundaries induces susceptibility.

VI.D. Nitrates and Nitrides

Several investigators have shown that the susceptibility of irons and steels in nitrate solutions is influenced by carbon, nitrogen, and substitutional alloy additions. Long and Uhlig (1965) found that oxygen in iron at a concentration as high as 0.07% does not have an appreciable effect on susceptibility to stress-corrosion cracking in boiling nitrate solutions. Uhlig and Sava (1963) showed that interstitial carbon or nitrogen in iron is necessary for susceptibility in boiling nitrate solutions. The critical minimum concentration of carbon in iron necessary for stress-corrosion cracking is equivalent to the solubility of carbon in ferrite at 200-350°C and the critical amount of nitrogen in iron for SCC attack at 150-250°C (Uhlig and Sava, 1963). Uhlig and Sava felt that reduction of both carbon and nitrogen contents to 0.001% or less promotes immunity from SCC when either water-quenched from 925°C or cold-rolled to 70-85% reduction in thickness. Long and Uhlig (1965) found the carbon content in steels water quenched from 925°C to be 0.002%. This is supported by Asphahani and Uhlig (1976). They do mention that the value of 0.002% carbon is probably influenced by unknown small amounts of nitrogen. They also found the critical carbon content for susceptibility to be 0.001% for heat-treated iron. Parkins (1952) states that cracking is associated with the presence of cementite in the grain boundaries of the ferrite. Parkins (1952) and Flis (1973) found that high carbon steels show resistance to stress-corrosion cracking by the decreased precipitation of cementite along grain boundaries and lower segregation of carbon atoms, and also by the enhanced deposition of protective magnetite on the corroding surface.

It is reported (Parkins, 1952; Uhlig, et al., 1974; Uhlig and Sava, 1963; and Asphahani and Uhlig, 1976) that cold rolled steel in the as received condition, which is probably severely cold worked, is resistant to stress-corrosion cracking. Uhlig, et al. (1974) found that cold rolling specimens to greater than about 50% reduction of thickness resulted in resistance to stress-corrosion cracking of all steels, except for a 0.0035% carbon alloy. Iron containing still less carbon tended to be resistant whether or not cold rolled. Imperfections generated by cold work probably attract carbon atoms from grain boundaries where carbon is necessary for susceptibility. Alternately, Asphahani and Uhlig (1976) found that severely cold rolled 1% nickel steels containing more than 0.12% carbon failed in a boiling nitrate solution by IGSCC.

Long and Lockington (1967) found that if substitutional alloy additions are of a carbide forming nature, then there will be a strong trend towards immunity. All of the 1%

chromium steels tested by Asphahani and Uhlig failed after being water-quenched, including those containing 0.005% or less carbon, which were resistant when not heat-treated. Several investigators (Uhlig et al., 1974; Long and Lockington, 1971; and Parkins and Brown, 1959), however, found that the addition of 1% chromium, a carbide and nitride forming element, resulted in resistance to stress-corrosion cracking in pure iron-based alloys by immobilizing C and N as carbides and nitrides, thus keeping C and N from the grain boundaries. They felt that increasing the chromium content would confer additional resistance to SCC in these alloys. Uhlig, et al. (1974) found that additions of molybdenum also imparts resistance to SCC. Long and Lockington (1971) also found that additions of the graphitizing element, nickel, at 1% had little effect on the time-to-failure. They attributed this to the fact that nickel has a solution strengthening effect that hinders movement of dislocations. Alternately, Uhlig, et al. (1974) found that additions of nickel up to 2% increases susceptibility; larger additions were found to decrease susceptibility. Parkins and Brown (1959) found that copper additions in mild steels may increase susceptibility towards stress-corrosion cracking in nitrate solutions.

For high-strength steels, according to Asphahani and Uhlig (1976), alloy composition is less important when water is the environmental medium. In this situation, hardness alone is the factor that apparently determines failure by stress-corrosion cracking, the critical hardness level for susceptibility being R_c 40 to 42, whether the steel is commercial 4140 or relatively pure low alloy steel containing nickel and chromium.

VI.E. Chlorides

Miyasaka and Ogure (1986) examined the influence of applied stress, alloying elements, temperature, NaCl concentration, dissolved oxygen concentration, and potential on the stress-corrosion cracking behavior of austenitic cast irons. This study yielded the following conclusions.

Austenitic cast irons have a high stress-corrosion cracking susceptibility in NaCl solutions, and hence, mitigating actions are needed when the austenitic cast irons are used in seawater or brine. As the applied stress was reduced, the time to failure was exponentially prolonged. ASTM A439 Type D2 austenitic cast iron has a higher stress-corrosion cracking resistance than ASTM A436 Type 2 austenitic cast iron and has a stress-corrosion cracking life more than 10 times as long as that of ASTM A436 Type 2 at the same stress. The stress-corrosion cracking mechanism of austenitic cast irons is of the active path corrosion cracking type, where stress-corrosion cracking is accelerated by anodic polarization but is decelerated by cathodic polarization. Even if the concentrations of nickel, silicon, and manganese of the austenitic cast irons are changed, stress-corrosion cracking can take place. However, an increase in nickel content leads to improved resistance to stress-corrosion cracking. At temperatures ranging from 25 to 52°C, in 7% NaCl solution, the higher the temperature, the shorter the time to failure. Three percent and 7% NaCl solutions have substantial difference in the time to failure. The resistance to stress-corrosion cracking is greatly increased by reduction in the dissolved oxygen concentration. Resistance to stress-corrosion cracking is higher in deaerated brine than in seawater. The most effective method for protection of austenitic cast irons from stress-corrosion cracking is cathodic protection. Other effective methods are reduction in the stress, increase in the nickel concentration of the material, and reduction in the dissolved oxygen of the environment.

Poulson and Parkins (1973) found that nickel in ferritic steels has the opposite effect on stress-corrosion cracking in $MgCl_2$. They found that nickel free steel was immune to stress-corrosion in $MgCl_2$. They also found that the addition of only 1% nickel to the steel promoted susceptibility, and the maximum susceptibility is at about 2% Ni.

Carter (1969) examined the effect of silicon on the stress-corrosion resistance of low-alloy, high-strength steels with 0.09% to 2.15% silicon contents. Carter established the stress-corrosion resistance of AISI 4340 steel for two tensile strength ranges (280 to 300 ksi and 230 to 240 ksi) in 3.5% sodium chloride solution. This study found that silicon additions of up to 2.15% to AISI 4340 steel did not increase the threshold stress parameter K_{ISCC} in the 280 to 300 ksi tensile strength range. He did, however, find that the stress-corrosion crack velocity was significantly retarded when the silicon content exceeded 1.5%. Carter also found that tempering to the 230 to 240 ksi tensile strength range improved K_{ISCC}. The crack velocity was found to be directly proportional to stress intensity in the 280 to 300 ksi tensile strength range and constant over a wide range of stress intensity for the tensile strength range of 230 to 240 ksi. Reducing the sulfur and phosphorus content of low-alloy, high-strength steels was found to enhance the crack growth resistance.

VI.F. Inclusions

In an effort to determine the role of inclusions in environmental cracking in an acid sulfate solution, Lees (1977) exposed stressed cold-worked C-Mn steel to pH 4 $Al_2(SO_4)_3/K_2SO_4$ solution containing 10^{-4} to 10^{-3} mol/liter H_2S. This study found that these conditions produce stable, discontinuous, transgranular crack growth in the steel. In addition, crack propagation rates were found to be an order of magnitude higher when inclusion stringers were aligned normal to the applied stress in the plane of the crack than when stringers are parallel to the crack advancing stage. Cracking was found to occur without H_2S additions in pH 3 solution owing to the generation of 10^{-3} moles/liter H_2S by chemical dissolution of sulfide inclusions. Further, he found that inclusions act as (i) local sites for H_2S generation, (ii) cathodic sites for H_3O^+ reduction, and (iii) low energy paths for H diffusion in the material.

Bernstein and Thompson (1976) conclude that for the entire range of iron-based alloys, control of susceptibility appears possible by modifications of composition, heat treatment, and structure, without sacrificing strength and toughness. The similarity in behavior of most iron alloys in a wide variety of environments and electrochemical conditions supports the occurrence of an often-dominant hydrogen component to the fracture process.

VII. Microbiologically Influenced Corrosion

This section provides a brief review of microbiologically-influenced corrosion (MIC) issues related to carbon steels in environments relevant to the Yucca Mountain site. A more detailed evaluation of this topic will be the subject of a subsequent degradation modes survey.

Bacteria associated with microbiologically-influenced corrosion are sulfate reducing bacteria, iron and manganese bacteria, and sulfur oxidizing bacteria (Licina, 1988). The sulfate-reducing bacteria, Desulfovibrio and Desulfomaculum, are the most widely distributed and economically important organisms associated with corrosion. They reduce sulfate to sulfide and are anaerobic, although they may be found in virtually all soils, and are capable of survival in aerated water. The survival of these rod-shaped bacteria depends upon the presence of oxygen-scavengers to achieve anaerobic conditions. The sulfur reducing bacteria attack cast iron, carbon steels, low-alloy steels, and stainless steels and can contribute to corrosion of high nickel alloys and copper alloys. These bacteria can produce massive deposits on and deep pitting of carbon steels.

Iron and manganese bacteria include Pseudomonas, Gallionella, Crenothrix, Leptothrix, Clonothrix, and Sphaerotilus. Pseudomonas are aerobic slime-formers that are generally found in corrosion deposits as relatively thin uniform films. Their main involvement in the corrosion process is to scavenge oxygen and harbor other species such as the sulfate-reducing bacteria. They also reduce Fe^{+3} to Fe^{+2} and in doing so may expose active metal and thereby increase corrosion. Gallionella are obligately aerobic iron bacteria that exude hydrated metal oxide. Their wastes are high in iron and manganese, and they can concentrate chloride. Microbiologically-influenced corrosion due to these bacteria is typically manifested by pitting of carbon and stainless steels causing hemispherical deposits on carbon steel and relatively deep pits in stainless steels.

Crenothrix, Leptothrix, Clonothrix, and Sphaerotilus are aerobic bacteria whose main effect on corrosion is to oxidize soluble iron to insoluble $Fe(OH)_3$. Crenothrix are more often found in the United States, while Leptothrix are more common to the United Kingdom, and Sphaerotilus are particularly abundant in heavily polluted water. Crenothrix, Leptothrix, Clonothrix, and Sphaerotilus generally form low, hemispherical tubercles on carbon steels with oxygen-depleted conditions beneath the tubercle. Many of these species tend to concentrate chloride and manganese as well leading to high local concentrations of chloride in oxygen-deficient areas. They are prime contributors to pitting and general corrosion in carbon steels. The influence of iron oxidizing bacteria can be two-fold:

1) The aerobic bacteria form a colony and the oxygen concentration is depleted such that a differential aeration cell is produced. Localized corrosion may then proceed at the localized anodic region.
2) As Fe^{+2} is oxidized to less soluble Fe^{+3}, the electrochemical potential at the surface is shifted in the noble direction such that the pitting potential is exceeded and pits initiate.

The iron-oxidizing bacteria may further assist in the autocatalytic process by oxidizing ferrous ion to the ferric state, thus keeping the potential in the pitting regime.

Iron bacteria contribute to corrosion by concentrating aggressive ions, depositing iron from solution, or even assisting in dissolution by reducing ferric ions to the ferrous state (e.g. Pseudomonas). These bacteria contribute to corrosion by creating differential aeration cells under which deposits or slime masses form. These slime masses harbor sulfate reducing bacteria. This results in the production of anaerobic conditions for the sulfate reducing bacteria and provides physical protection from fluid forces and biocides.

The sulfur-oxidizing bacteria, Thiobacillus, Ferrobacillus, Thiothrix, and Beggiatoa, produce sulfuric acid in their metabolic process, which is a potential but seldom reported, source of metallic corrosion. Thiobacillus and Ferrobacillus are strictly aerobic and obligately autotrophic. Thiobacillus oxidize sulfide to sulfate and form H_2SO_4; some species can generate extremely acidic conditions (pH less than 1). They also oxidize ferrous ions to the ferric state and often coexist with sulfate-reducing bacteria. They do not, however, appear to be a major contributor to corrosion of metals. Thiothrix and Beggiatoa are aerobic bacteria that oxidize H_2S to elemental sulfur. Their major contribution to corrosion processes is to form a slime in which sulfate reducing bacteria may thrive. These same sulfur-oxidizing bacteria frequently contribute to the failure of concrete (Licina, 1988).

Microbiologically-influenced corrosion effects carbon steel by random pitting and general corrosion. Pitting under tubercles is common and exhibits no preference for a particular microstructural constituent. Tubercle formation appears to be a necessary, although not sufficient, condition for pitting. Pitting results as differential aeration and concentration cells are created by mechanical (i.e., tubercle formation) and chemical influences (i.e., oxygen depletion, concentration of chlorides, or reduction of sulfate and the resultant effects on cathodic depolarization) of microbiological activity. Mechanical crevices resulting from design or fabrication produce environments conducive to both growth of bacteria and localized abiotic corrosion. The result is often enhanced microbiologically influenced corrosion effects.

VIII. Weldability and Joining

This section provides a brief review of weldability and joining issues related to carbon steels in environments relevant to the Yucca Mountain site. A more detailed evaluation of this topic will be the subject of a subsequent degradation modes survey.

The American Welding Society defines weldability as "the capacity of a material to be welded under the fabrication conditions imposed into a specific suitably designed structure and to perform satisfactorily in the intended service." The ASM *Metals Handbook* (1993) suggests that the best definition is that a weldable material can be welded so that the joint is equal in all respects to the base metal. Following is a brief overview of some factors affecting the weldability of low-alloy steels.

The ASM *Metals Handbook* (1993) describes the desirable weld joint as having uniform strength, ductility, notch toughness, fatigue strength, and corrosion resistance throughout the weld and adjacent material. Welds may experience two types of problems:

1. Problems of the weld metal deposit or heat-affected zone that occur in connection with or immediately following the welding operation, such as hot cracking, heat-affected zone cracking, hydrogen-induced cracking, etc.
2. Problems in the weld or adjacent to the weld that occur any time during service of the weldment. These can be any kind of defects that will reduce the efficiency of the weld joint under service conditions.

It is the objective of the welder to create a weld that will avoid the problems. Restraint, weld shape, excessive heat input, or material composition may contribute to hot cracking. Hot cracking can occur as a result of any of these factors but is much more likely if two or more factors are present.

Hydrogen-induced cracking occurs when the conditions outlined in 1-4 (below) occur simultaneously (Bailey, et al., 1993):

1. *Hydrogen is present to a sufficient degree*
   Hydrogen is always present. It is absorbed by the weld pool from the fluxes used in welding. Some hydrogen is transferred to the heat-affected zone by diffusion from the solidifying weld pool.
2. *Tensile stresses act on the weld*
   These are caused by the contracting of the metal during cooling and may be supplemented by other stress formed as a result of rigidity in the parts joined.
3. *A susceptible heat-affected zone microstructure is present*
   That part of the heat-affected zone that experiences a high enough temperature for the parent steel to transform rapidly from ferrite to austenite and back again produces microstructures which are usually harder and more susceptible to hydrogen embrittlement than other parts of the heat-affected zone. Hydrogen cracks, when present, are invariably found in these transformed regions.
4. *A low temperature is reached*
   The greatest risk of cracking occurs when temperatures near ambient are reached and cracking may thus take place several hours after the welding has been completed.

The ASM *Metals Handbook* (1993) cites the four main factors that affect heat-affected zone cracking as follows: 1) the thickness of the base metal and the type of weld, 2) the composition of the base metal, 3) the welding process and filler metal type, and 4) energy input and preheat temperature.

The two factors most important to weldability are hardenability and the susceptibility of hardened microstructures to cracking. Both are increased by using a higher carbon or higher alloy content in the base metal and weld metal. Different alloying elements affect susceptibility to cracking less than others. Therefore, the "carbon equivalent" of the base metal is important. "Carbon equivalent" is a measure of the ferrite stabilizing effects of alloying additions relative to a similar effect noted for a specific addition of carbon. There are many empirical expression for "carbon equivalent" depending on the steels composition . *Metals Handbook* (1993) cites a representative carbon equivalent formula as follows:

$$CE = \%C + \frac{\%Mn}{6} + \frac{\%Mo}{4} + \frac{\%Cr}{5} + \frac{\%Ni + \%Cu}{15} + \frac{\%P}{3} \qquad \text{(VIII-1)}$$

The International Institute of Welding carbon equivalent equation is (*Metals Handbook*, 1993):

$$CE = \%C + \frac{\%Mn}{6} + \frac{\%Mo + \%V + \%Cr}{5} + \frac{\%Ni + \%Cu}{15} \qquad \text{(VIII-2)}$$

Figure VIII-1 summarizes the carbon equivalent and weldability description of some steel families. This figure is based on equation VIII-1 and should be used cautiously for actual welding situations. Other expressions for carbon equivalent have been developed based on solution thermohydraulics and kinetic considerations to obtain better predictions of alloy behavior and weldability of low-carbon low-alloy steels.

The following is a summary of the behavior of low-carbon, low-alloy steels with respect to weldability based on carbon equivalence. Plain carbon steels that have carbon equivalents of not over 0.40% are considered readily weldable. This carbon equivalent can be increased up to 0.45% provided that the carbon does not exceed 0.22%, the phosphorus does not exceed 0.06%, and the steel is not over 1.91 cm thick.

Usually when the carbon equivalent exceeds 0.40% special controls are required. The low-hydrogen processes or filler metals should be employed. Higher heat input should be employed and preheat may be required. When the carbon equivalent exceeds 0.60%, low-hydrogen processes are required, and preheating is required if the thickness is in excess of 1.91cm.

Hardenability tends to be higher for faster cooling rates. The cooling rate depends on the mass of the metal, the welding process, the welding procedure, and preheat temperature. Greater energy input also results in decreased cooling rates. The energy input is calculated by

$$HE = E * I * \frac{60}{S} \tag{VIII-3}$$

where HE = energy input in joules per linear measure of weld (inches or centimeters),

E = the arc voltage in volts,

I = the welding current in amperes,

S = the travel speed in lineal measure per minute (in/min or cm/min)

Therefore, increasing amperes or voltage or decreasing travel speed results in a decrease in the cooling rate and thus, the hardness.

Low-carbon steels, with from 0.10 to 0.25% carbon and from 0.25 to 1.5% manganese, can be easily welded with any of the arc, gas, and resistance welding processes. Medium-carbon steels contain carbon ranges of 0.25 to 0.50% carbon and manganese ranges of 0.60 to 1.65%. With higher carbon and manganese the low-hydrogen type electrodes are recommended, particularly in thicker sections. Preheating the base metal to temperature in the range from 149-260°C may be required for the completion of a successful weld. Post-weld heating is often specified to allow enough time for hydrogen to diffuse out of the weld and heat affected zone and help reduce hardness that may have been caused by rapid cooling. These medium-carbon steels are very difficult to weld even when the above precautions are observed, as shown in Figure VIII-2 (Czyryca, 1980). This figure is a representation of the weldability of iron-base alloys as a function of chemical composition and carbon equivalent. The carbon equivalent in this figure is based on Equation VIII-2.

High-carbon steels, 0.50 to 1.03% carbon, can be welded by the same processes as low-carbon steels provided that low-hydrogen electrodes are employed and the base metal is preheated to between 204-316°C. Also, the welding should be followed by stress-relieving or annealing. Such post-weld heat treatment would be very difficult for large structure such as the proposed Multi-Purpose Container design for the Yucca Mountain repository. Free-machining low-alloy steels containing sulfur in amounts of 0.050% to 0.15% are difficult to weld. These levels of sulfur in the steel have a tendency to produce hot cracking in the weld. Weathering steels can be welded by all of the arc welding processes and by gas and resistance welding. However, special welding procedures should be used to maintain the corrosion resistance characteristic of the steel.

Ductile irons can be welded using gas metal arc welding, short arc welding, and oxyacetylene welding (Karsay, 1975). However, welding techniques for ductile irons were developed mainly for the purpose of repair. Distortion and cracking in the heat affected area of welded ductile iron are minimized by preheating to between 260-540°C. Also a postweld heat treatment may be necessary, depending on the structure of the weld deposit and the severity of stresses created. For this reason, the ductile cast irons that are to be heat treated need to be welded as-cast, followed by the heat treatment.

When welding malleable cast iron with a process that heats the base metal above 760°C, the metallurgical structure in part of the heat-affected zone is changed to white cast iron (Thielsch and Dawson, 1958). This is undesirable. It is necessary to either use a welding process that does not heat the malleable cast iron above 760°C, which is not possible for large structures or to perform a postweld heat treatment on the completed casting. This postheating involves heating the cast to between 675-735°C for eight hours followed by slow cooling.

Wrought iron can be welded easily by any of the commonly used processes (Aston and Story, 1939). The absence of the A_1 point in wrought iron is of importance in reducing internal strains due to welding and in eliminating air-hardening. However, stress relieving of welded structures is recommended at between 370-430°C.

IX. Conclusions

The oxidation and atmospheric corrosion data suggest that the addition of chromium provides the greatest improvement in oxidation resistance. Chromium-bearing cast irons are resistant to chloride environments and solutions containing strongly oxidizing constituents. Weathering steels, including relatively high carbon content and at least 0.04% Cu, appear to provide adequate resistance to oxidation under temperate conditions. However, data from long-term, high-temperature oxidation studies on weathering steels were not available.

From the data identified in the literature, it appears that the low alloy steels, plain carbon steels, cast steels, and cast irons corrode at similar rates in an aqueous environment. Alloys containing more than 12% Cr or more than 36% Ni corrode at a lower rate than plain carbon steels, but pitting may be worse. Short term tests indicate that an alloy of 9Cr-1Mo may result in increased corrosion resistance, however long term data are not available. Austenitic cast irons show the best corrosion resistance.

A ranking of the total corrosion performance of the materials investigated form most corrosion resistant to least corrosion resistant is provided below.

1. Austenitic Cast Iron
2. 12% Cr = 36% Ni = 9Cr-1Mo
3. Carbon Steel = Low Alloy Steels
4. Cast Iron

Since the materials to be employed in the Advanced Conceptual Design (ACD) waste package are considered to be corrosion allowance materials, the austenitic cast irons, high Cr steels, high Ni steels and the high Cr-Mo steels should not be considered as candidates for the outer containment barrier. Based upon the oxidation and corrosion data available for carbon steels, low alloy steels, and cast irons, a suitable list of candidate materials for a corrosion allowance outer barrier for an ACD waste package could include, A516, 2.25%Cr - 1%Mo Steel, and A27.

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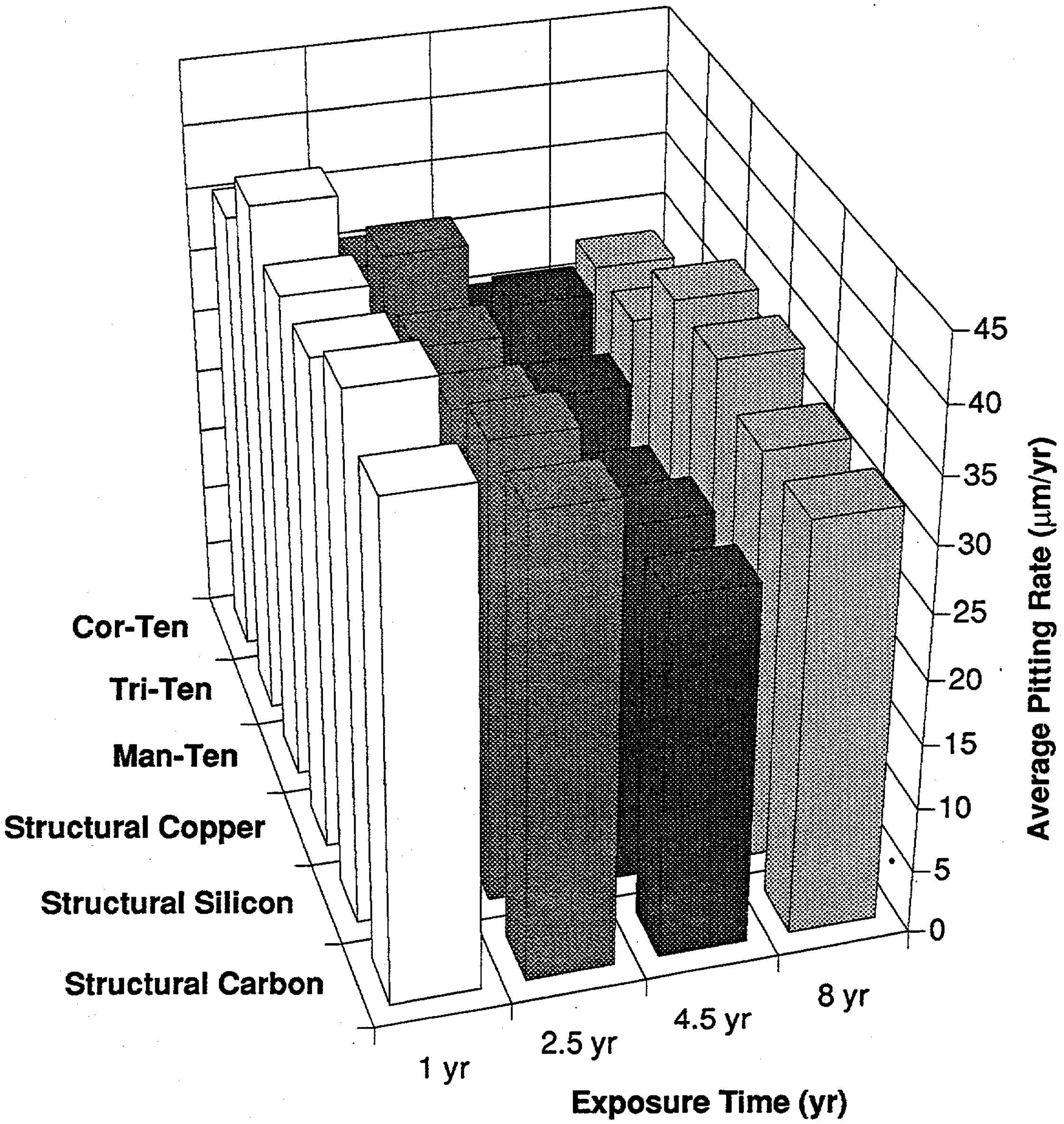

Figure III-8 Average pitting rate of structural steel in the Mississippi River (Larrabee, 1953).

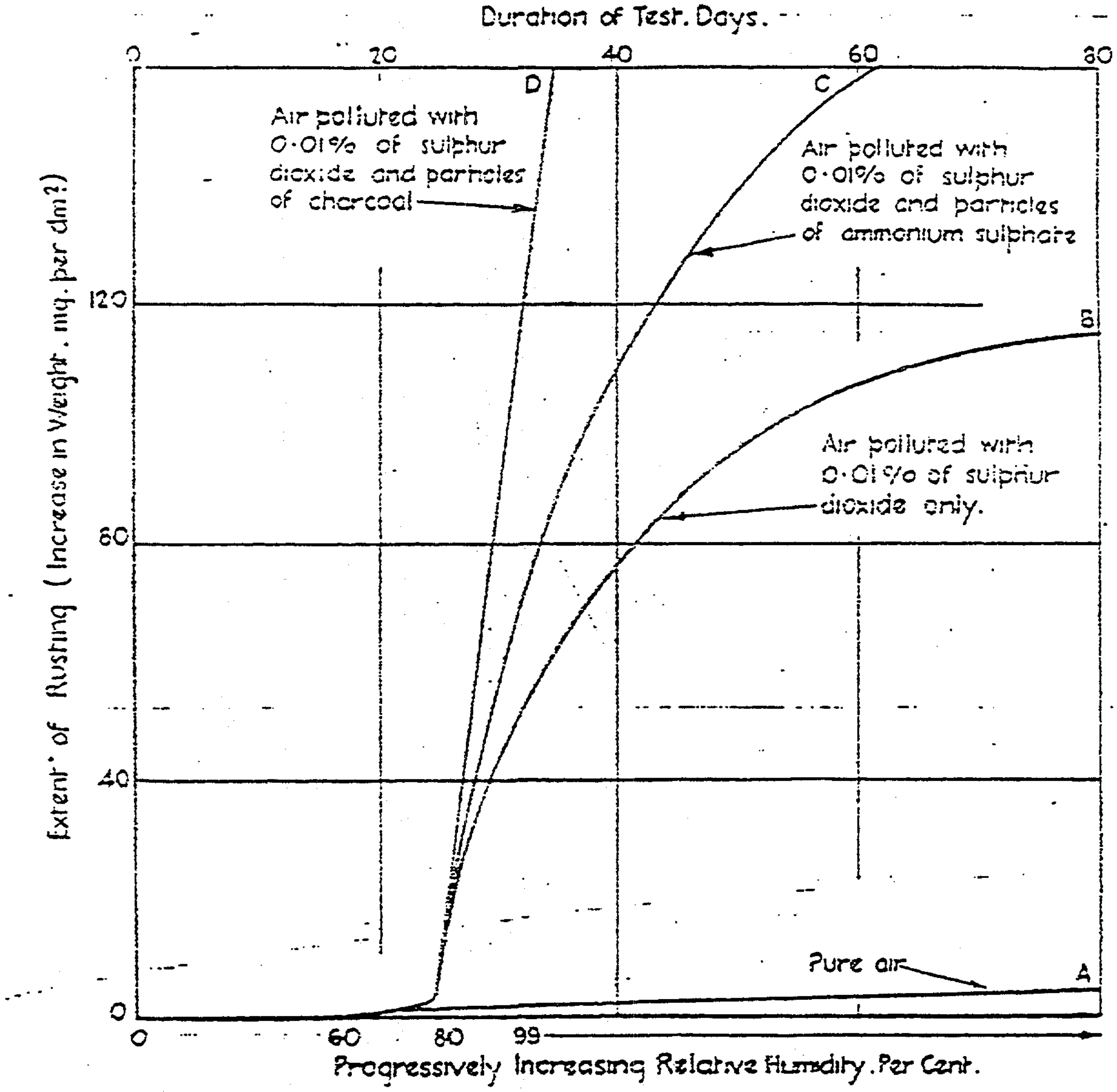

Figure II-1. Influence of humidity and atmospheric pollution on the rusting of mild steel (Vernon, 1949).

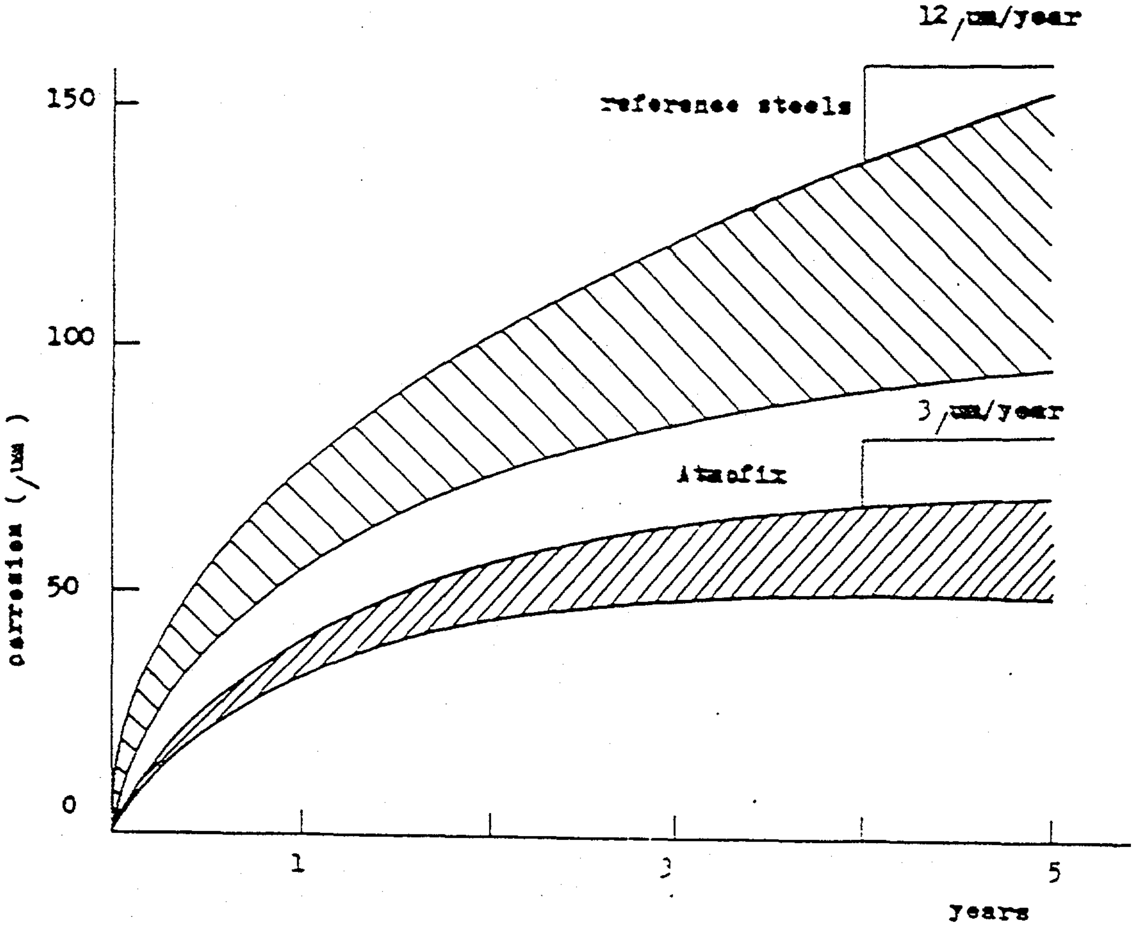

Figure II-2. Corrosion losses of Atmofix 52A and the reference steels in the outdoor urban and weakly polluted industrial atmospheres (limit of SO2 pollution is 90 mg/m2/day, Knotkova-Cermakova, Vlckova, and Honzak, 1982).

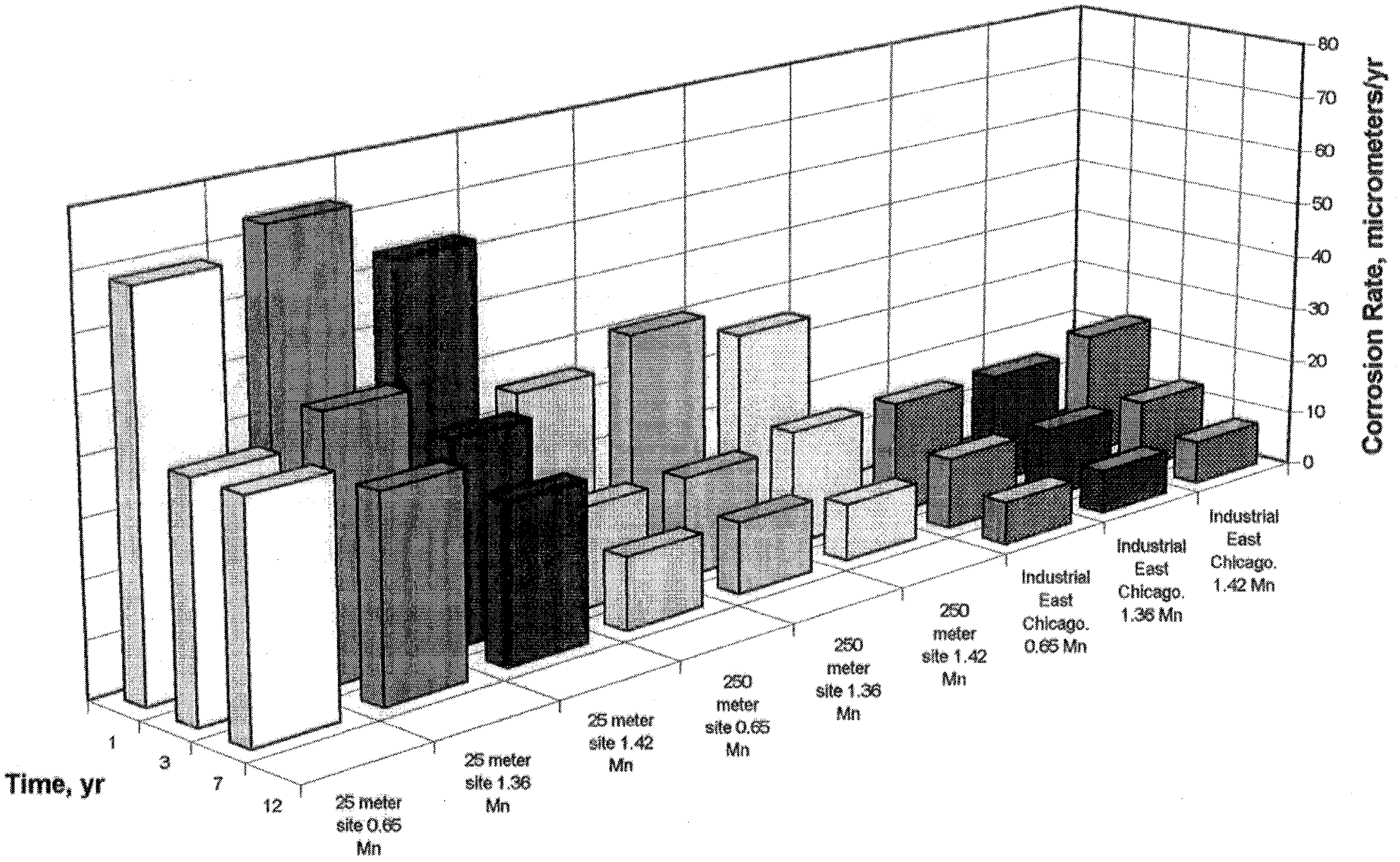

Figure II-3 Atmospheric corrosion rate of high-manganese cast steels (Briggs, 1968).

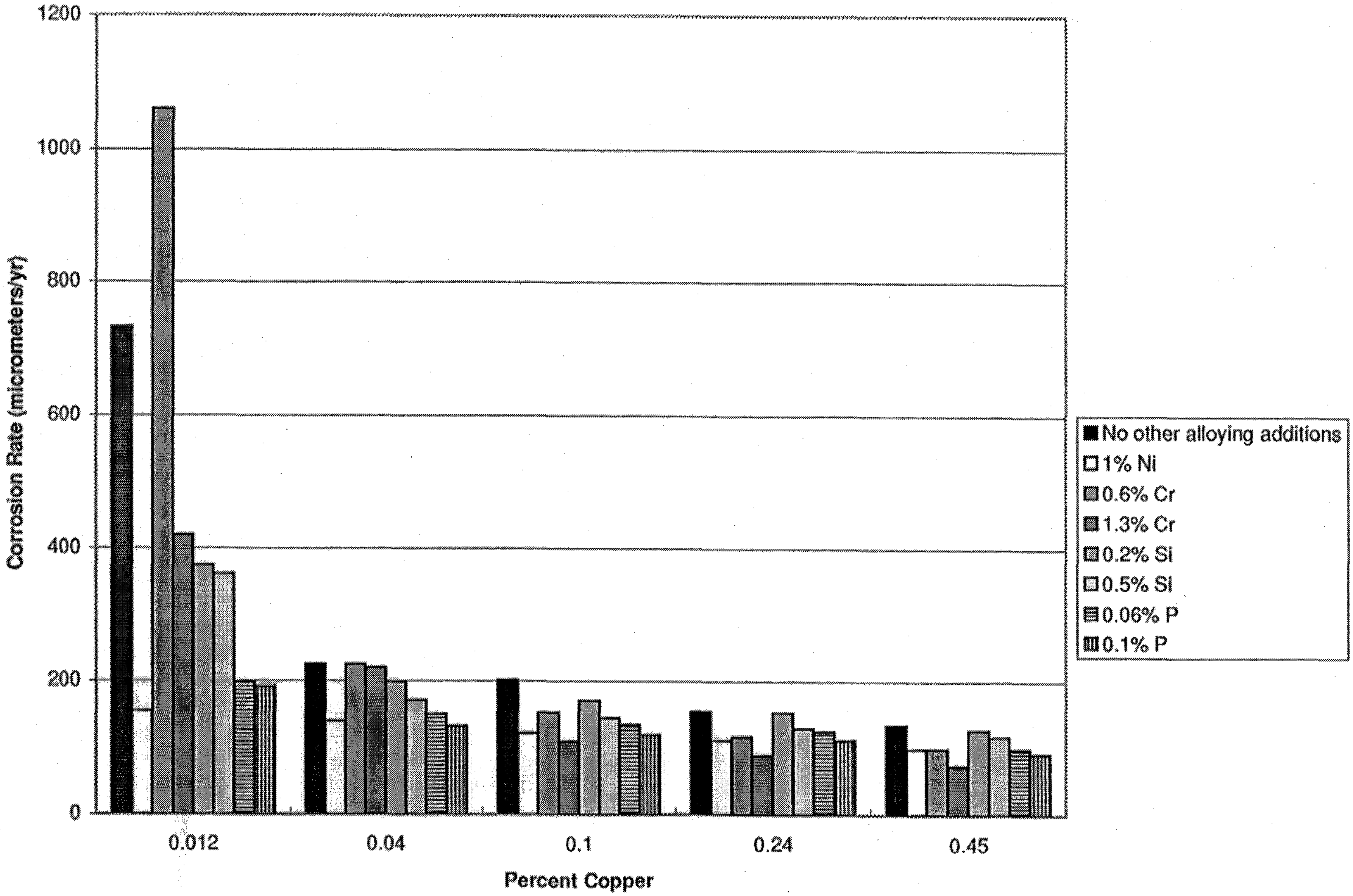

Figure II-4 Effect of single alloy additions and copper in an industrial environment (Larrabee and Coburn, 1961).

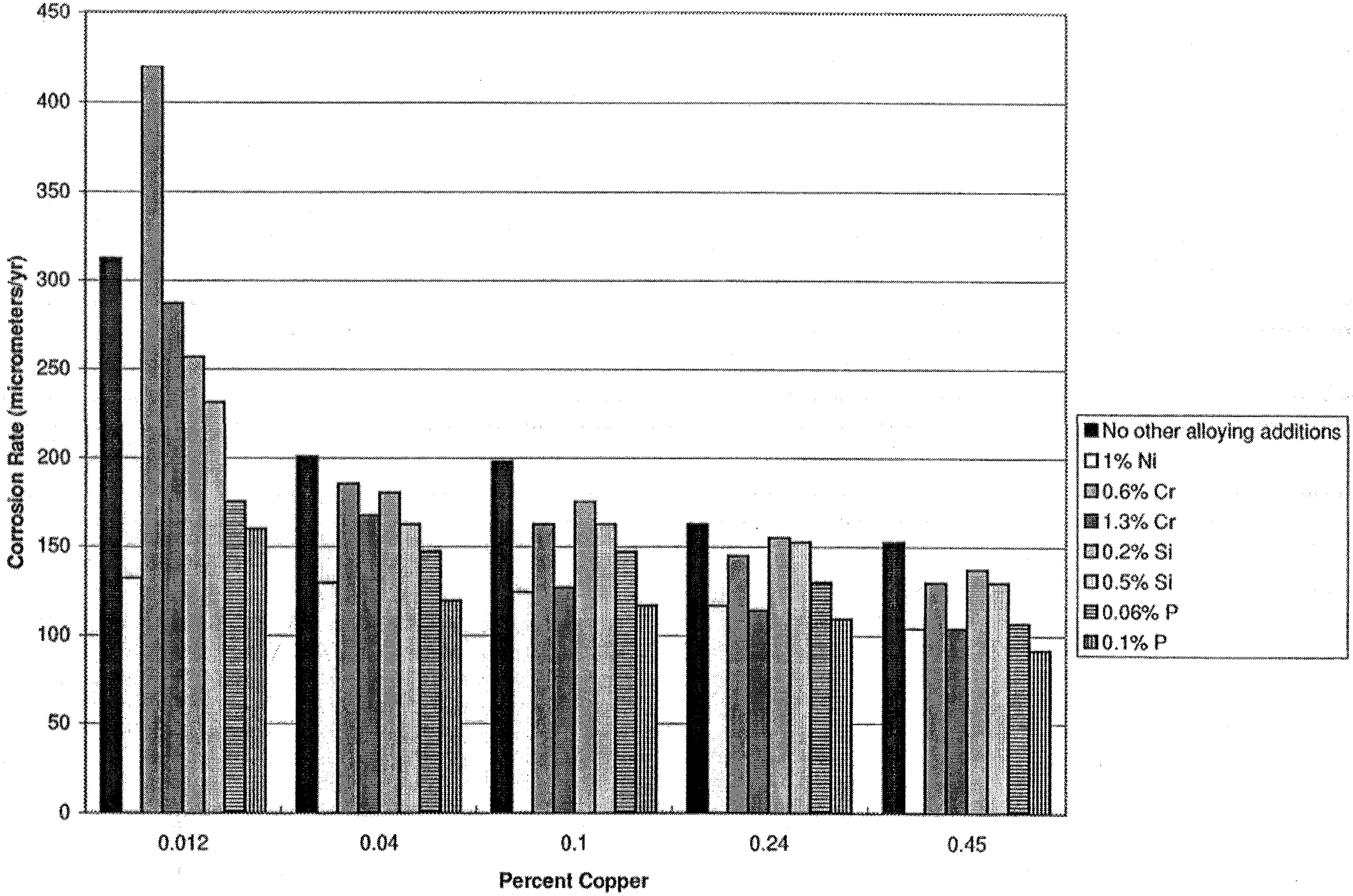

Figure II-5 Effect of single alloy additions and copper on corrosion in a semi-rural environment, (Larrabee and Coburn, 1961).

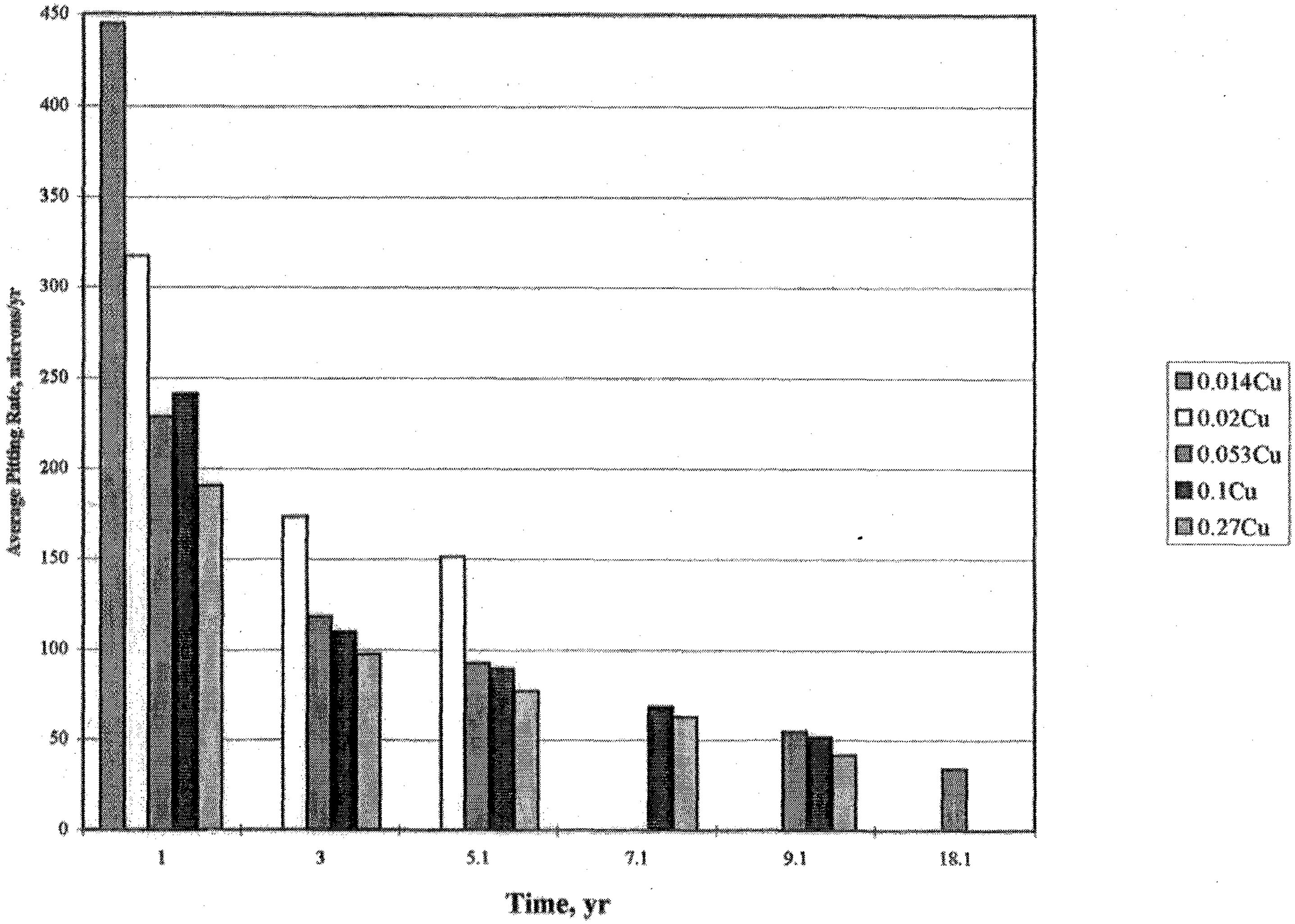

Figure II-6 Average pitting rate for high-purity plus copper iron exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

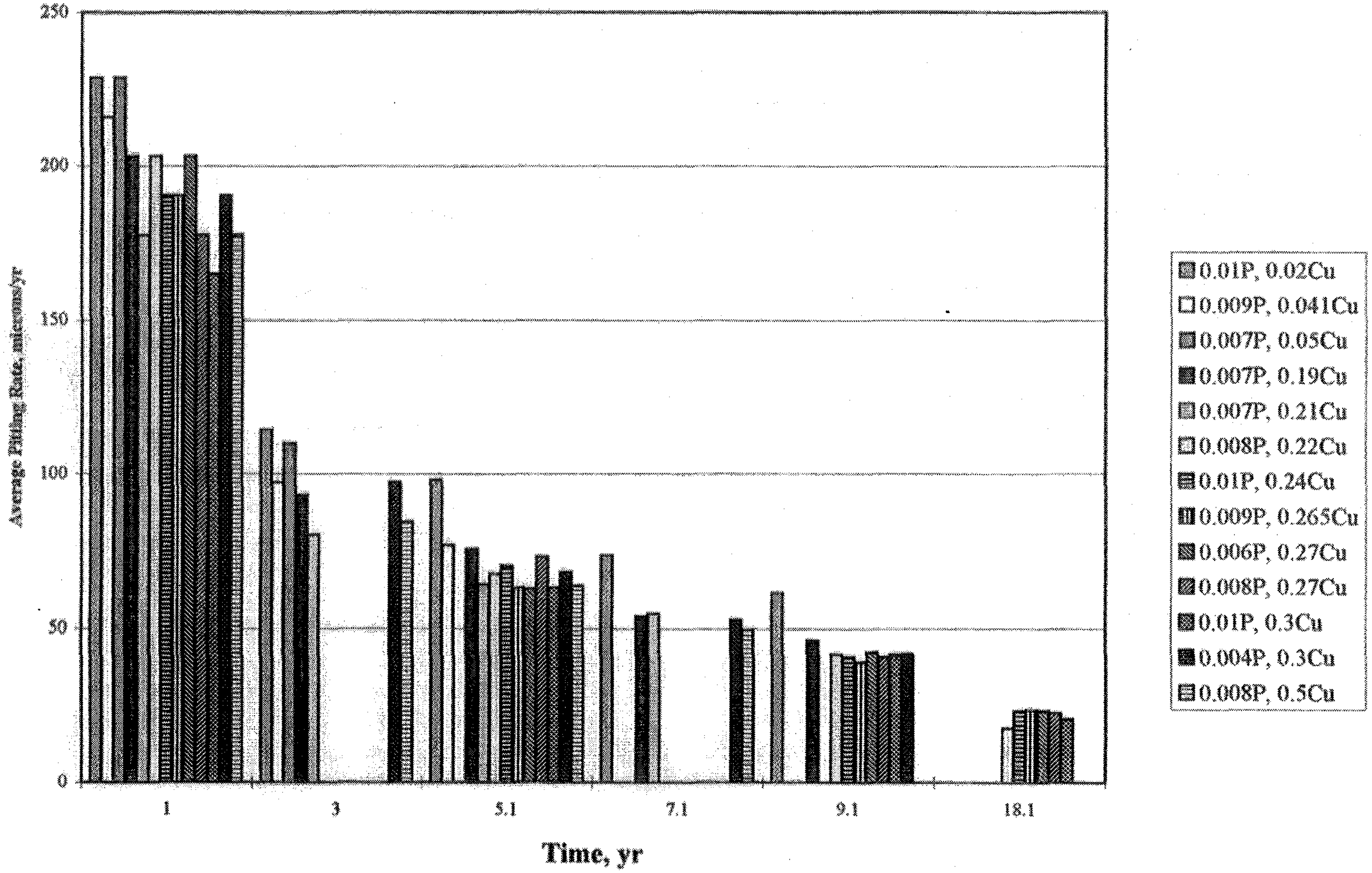

Figure II-7 Average pitting rate for low-phosphorus plus copper steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

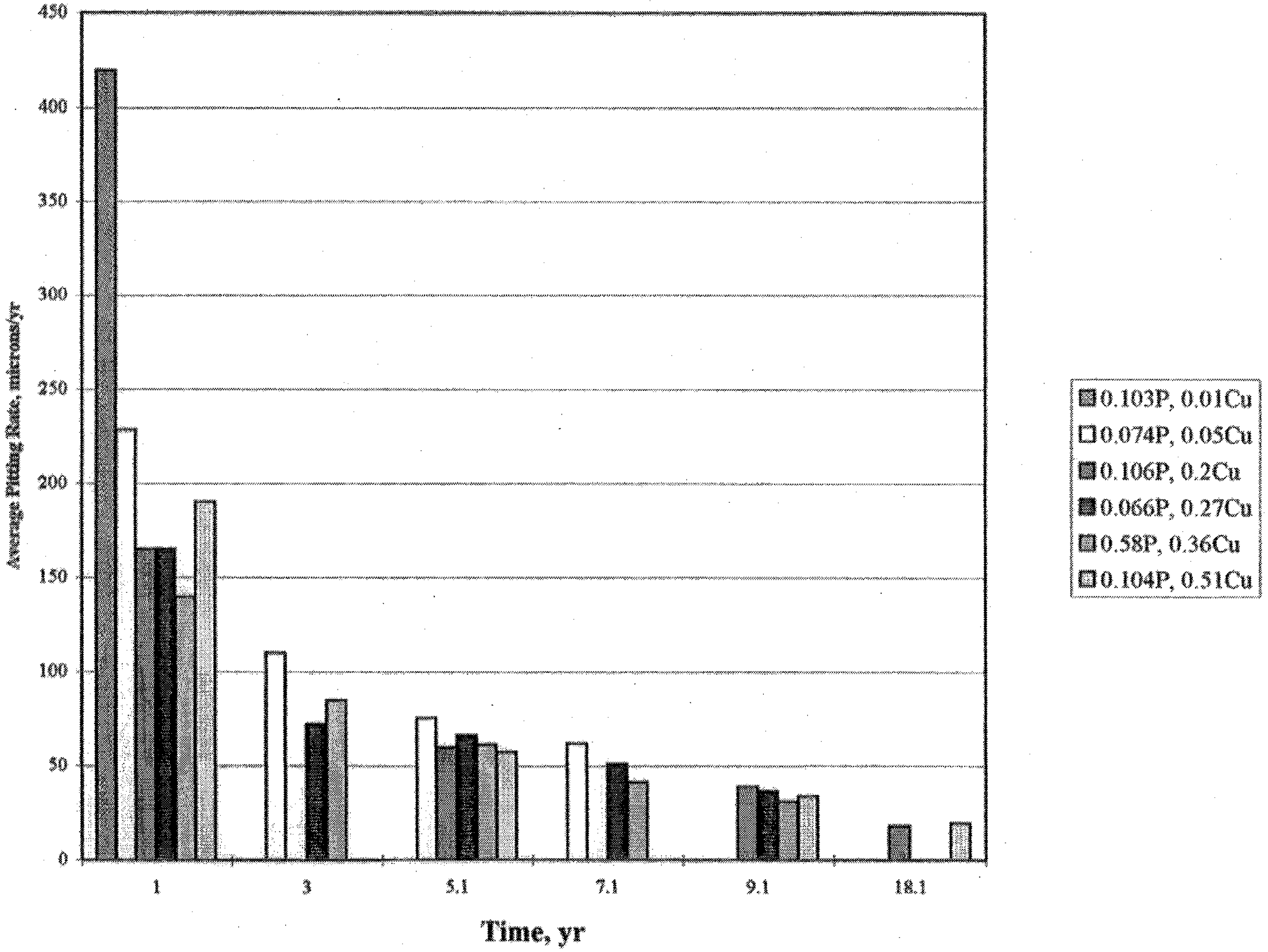

Figure II-8 Average pitting rate for high-phosphorus plus copper steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

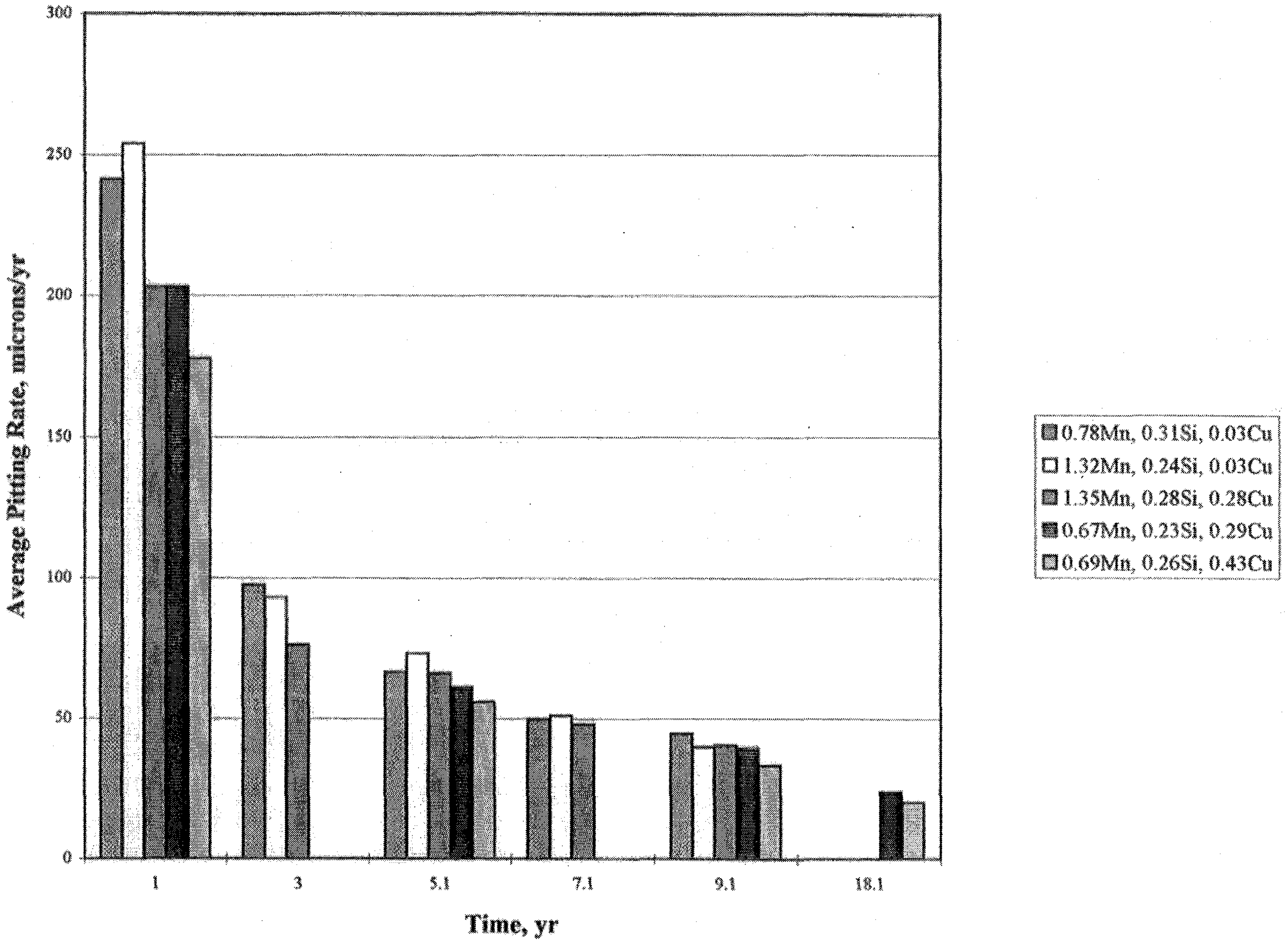

Figure II-9 Average pitting rate for high-manganese and silicon plus copper steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

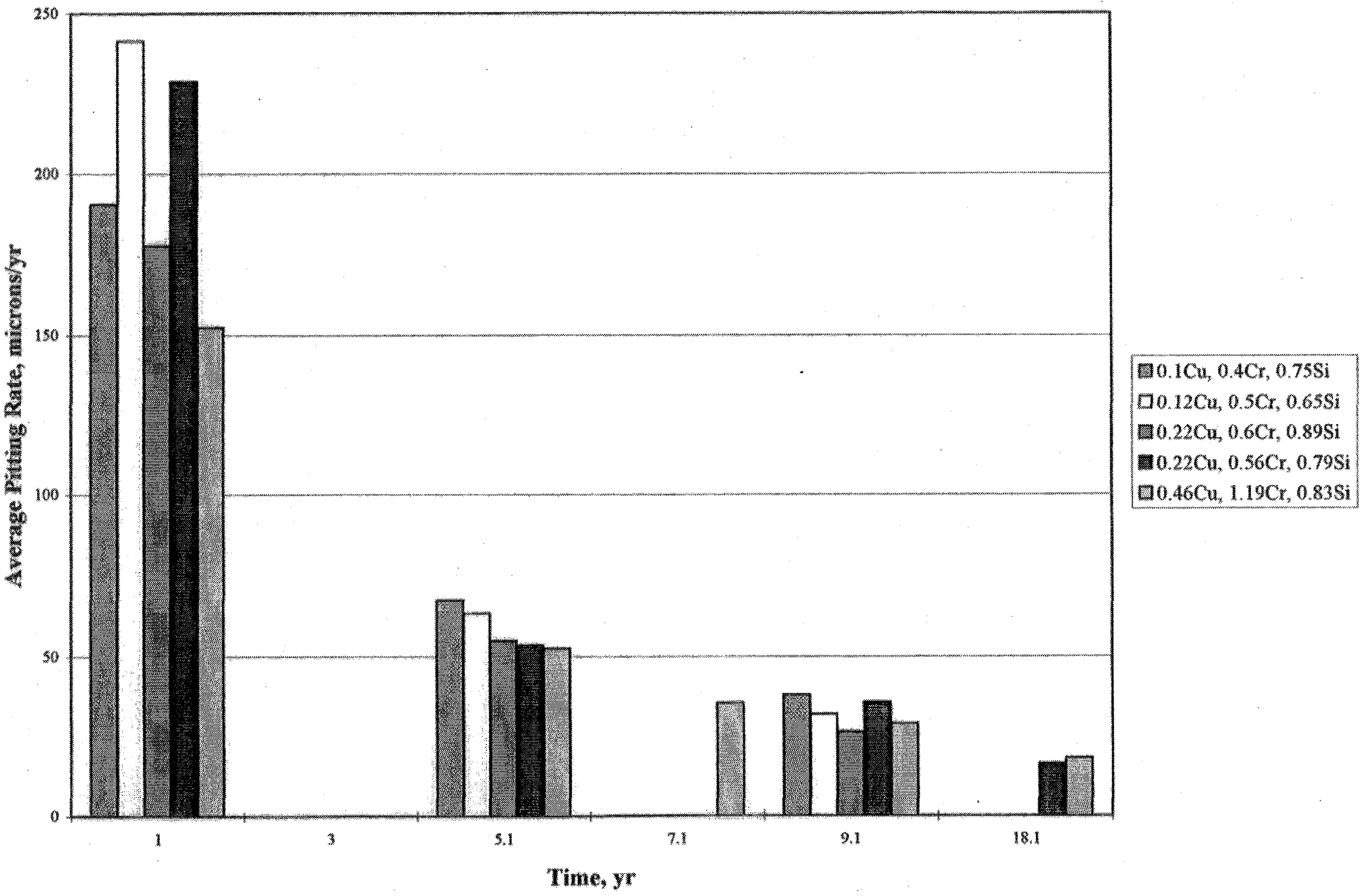

Figure II-10 Average pitting rate for copper plus chromium and silicon steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

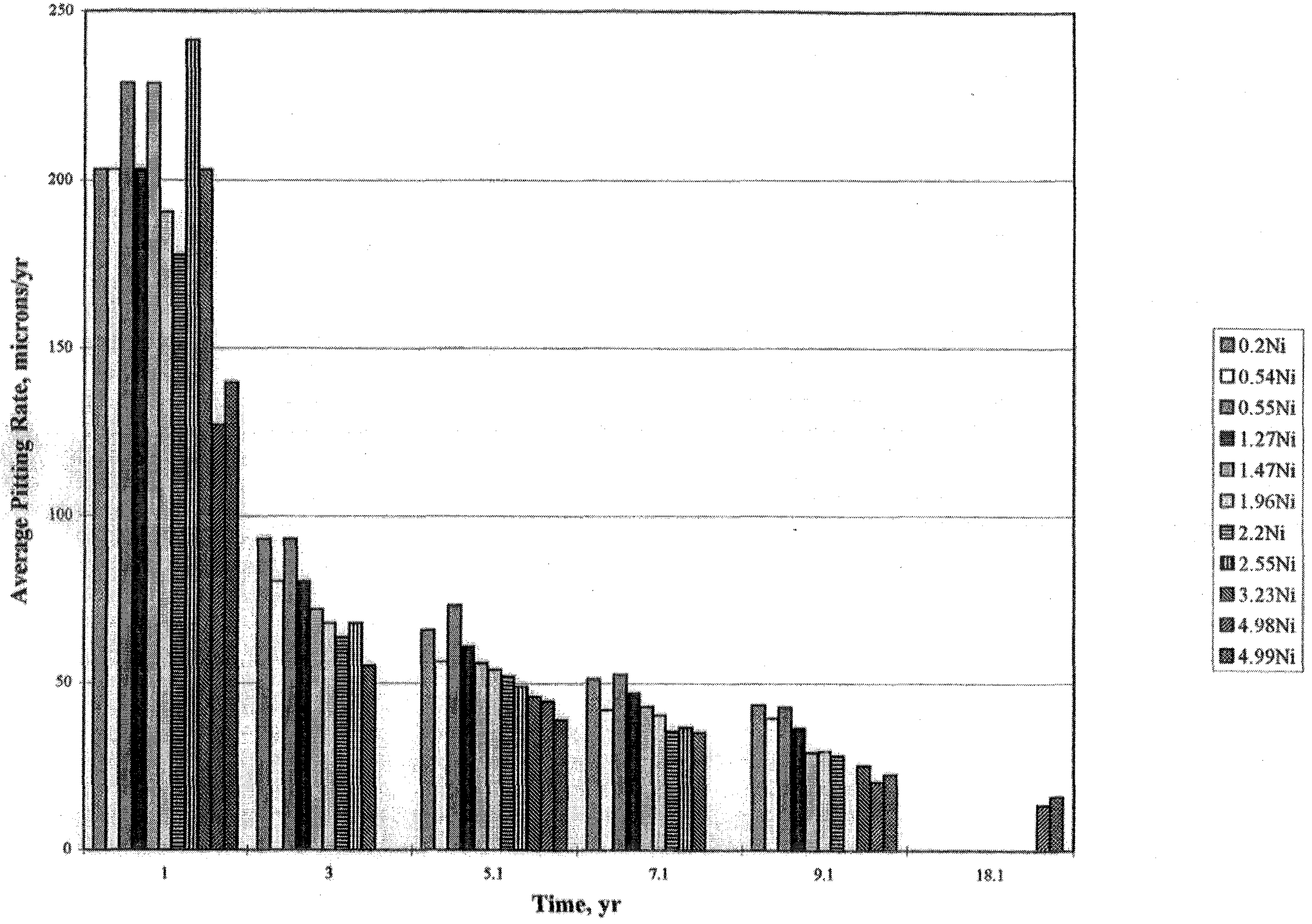

Figure II-12 Average pitting rate for nickel steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

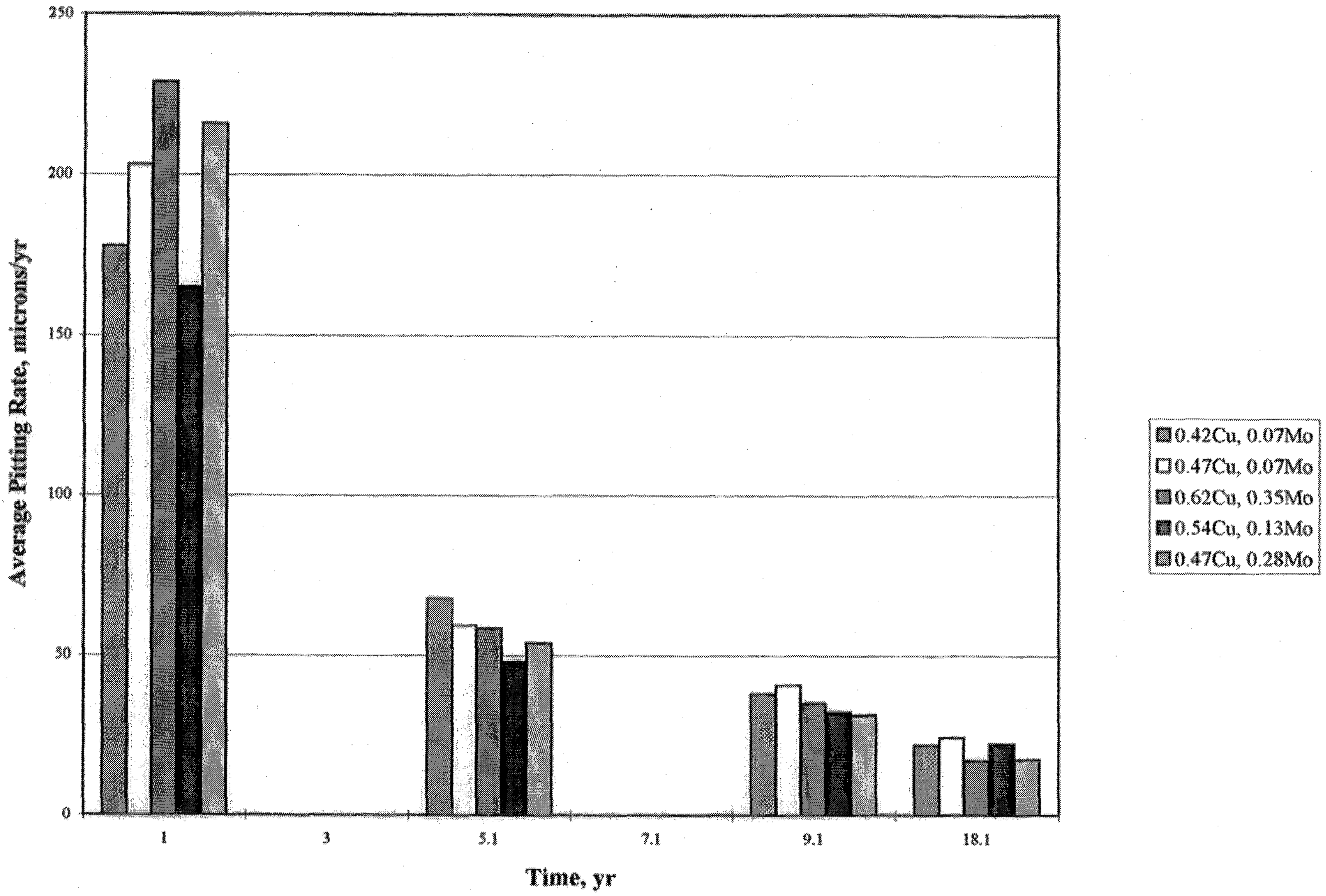

Figure II-11 Average pitting rate for copper plus molybdenum steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

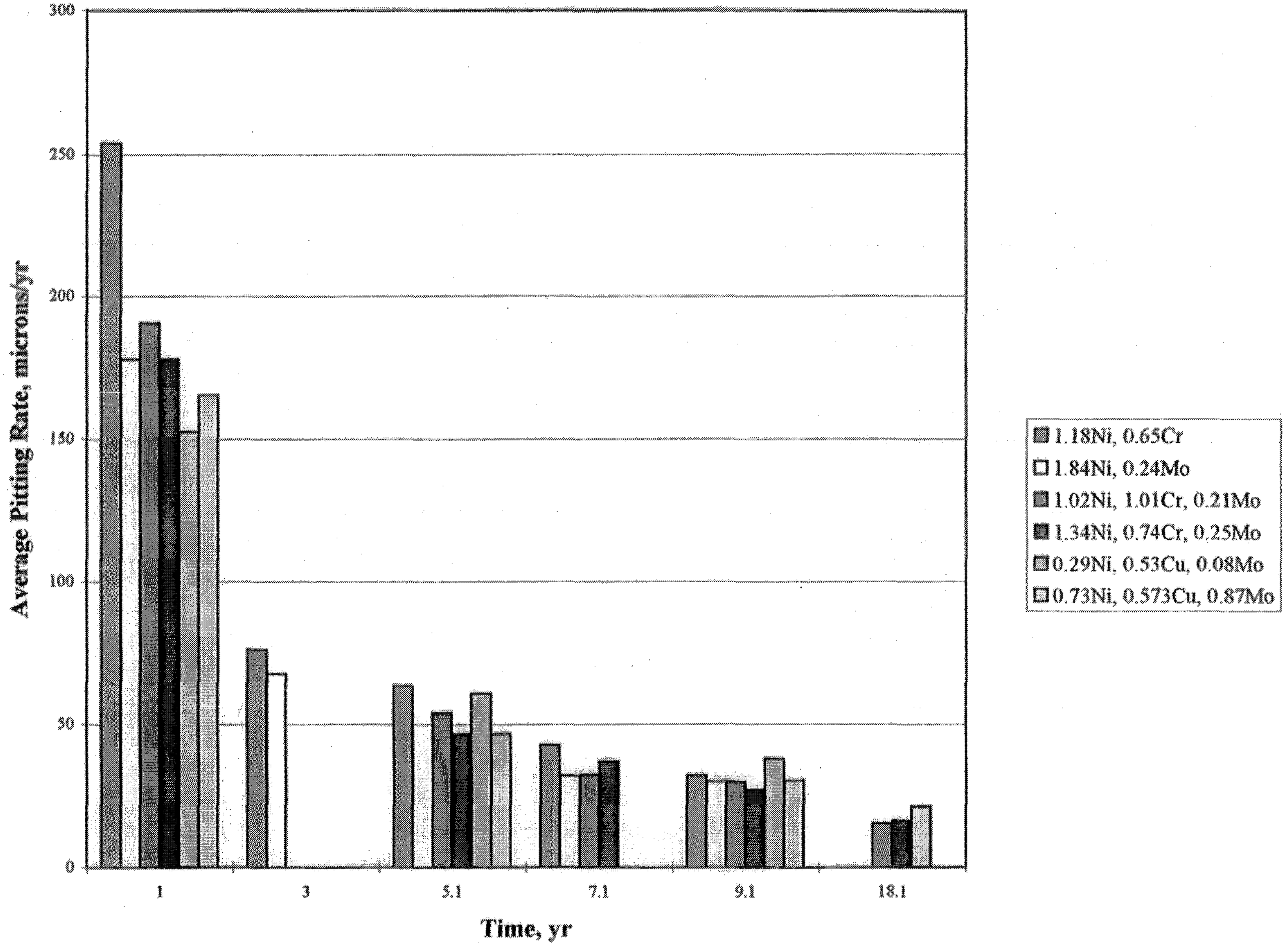

Figure II-13 Average pitting rate for nickel plus chromium and molybdenum steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

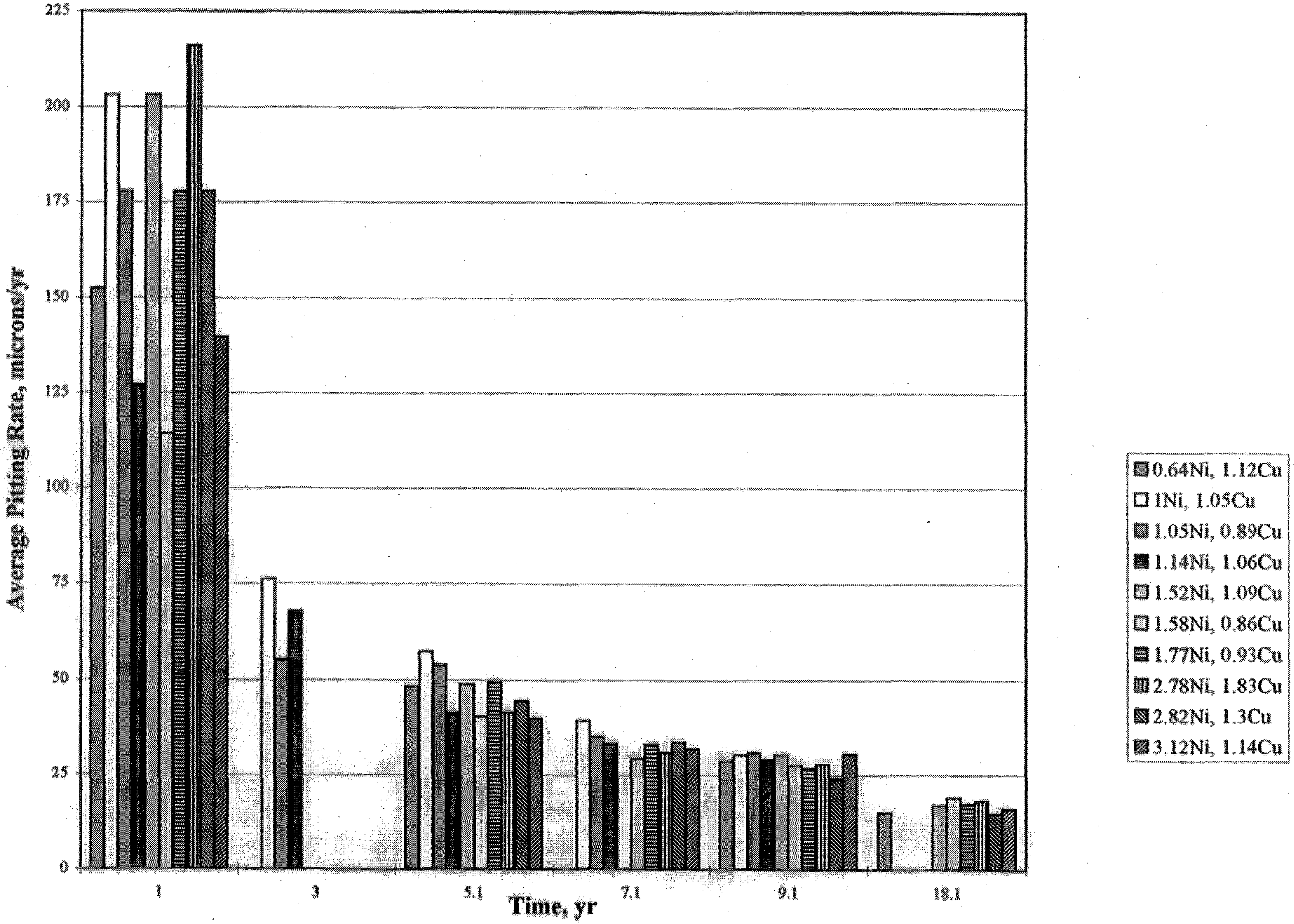

Figure II-14 Average pitting rate for nickel-copper steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

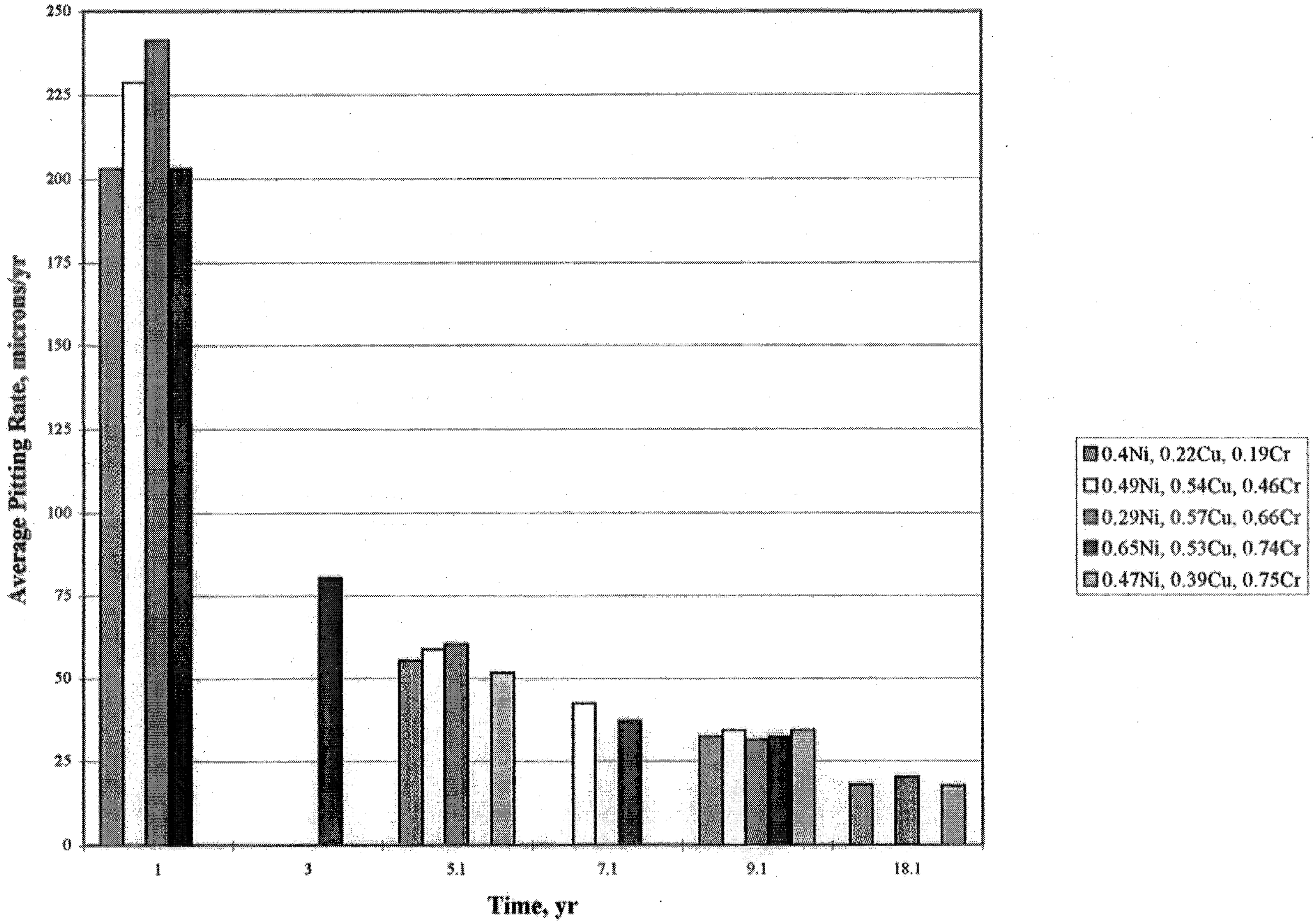

Figure II-15 Average pitting rate for nickel-copper plus chromium steel exposed to an industrial environment based on 4 skyward and 4 groundward deepest pits (Copson, 1945; 1948; 1952; 1960).

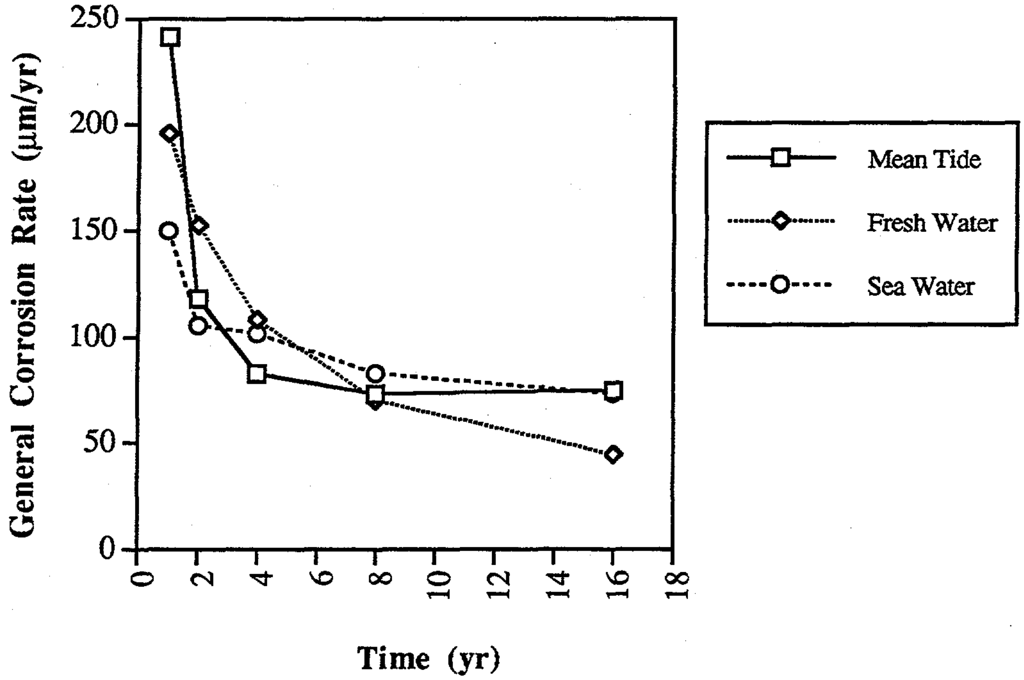

Figure III-1 General corrosion rate as a function of time for carbon steel exposed to various waters in the Panama Canal Zone (Southwell & Alexander, 1970).

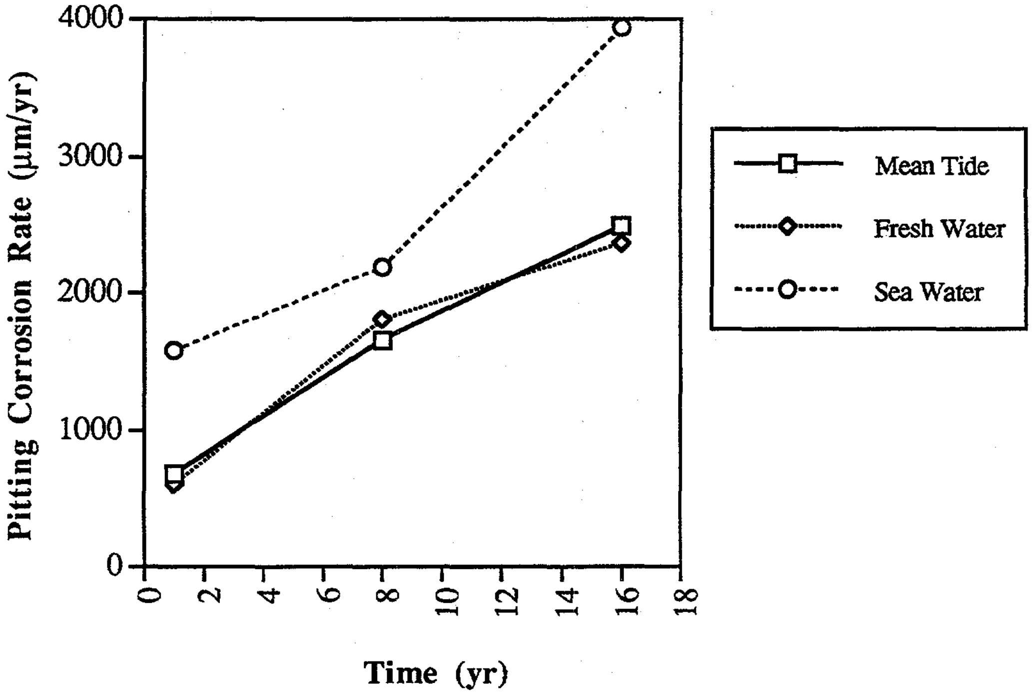

Figure III-2 Pitting corrosion as a function of time for carbon steel exposed to various waters in the Panama Canal Zone (Southwell & Alexander, 1970).

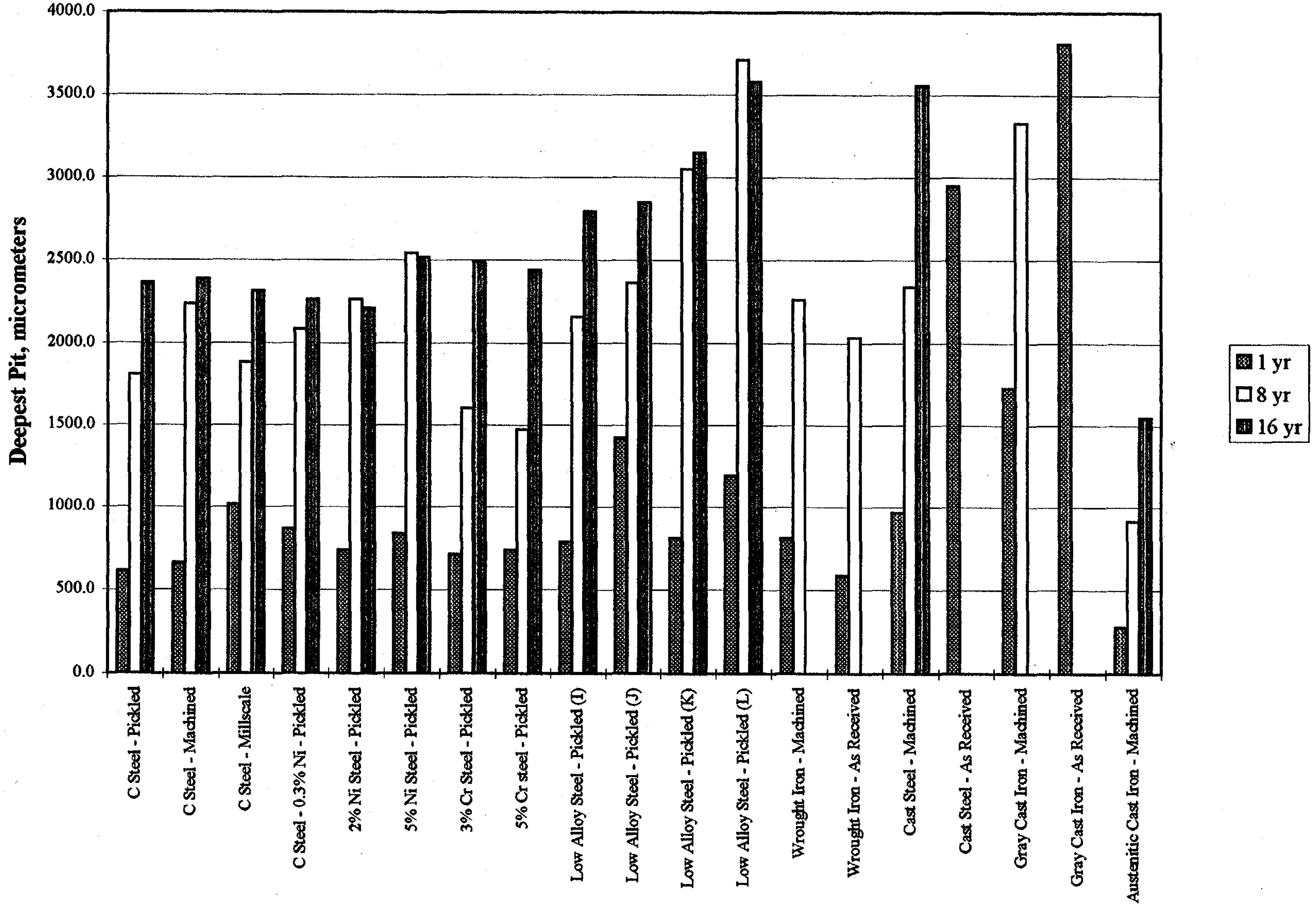

Figure III-3 Maximum pit depth of continuously immersed structural ferrous metals in fresh water - Gatun Lake, Panama Canal Zone (Southwell Alexander, 1970).

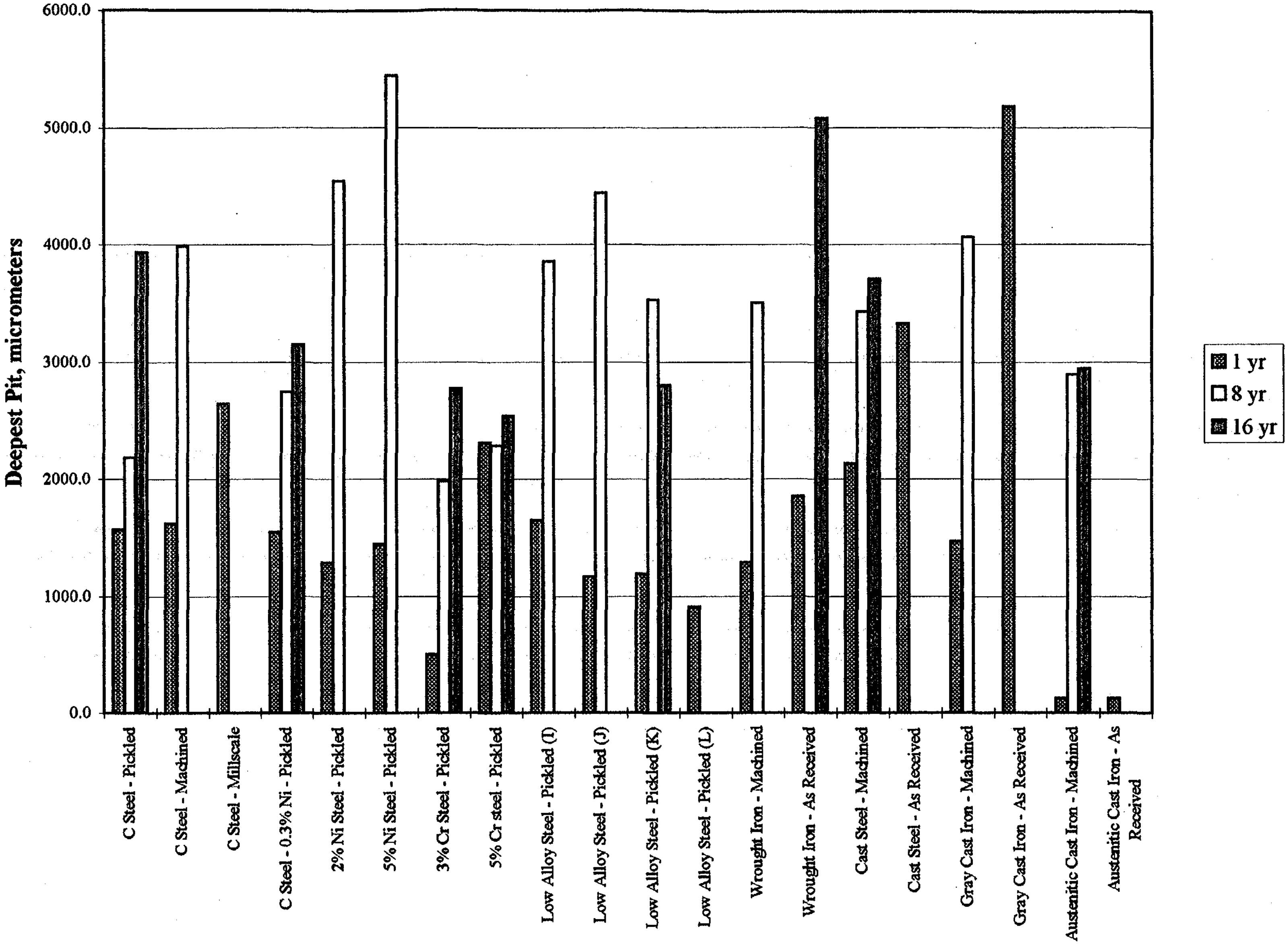

Figure III-4 Maximum pit depth of structural ferrous metals continuously immersed in sea water - Pacific Ocean, Panama Canal Zone (Southwell Alexander, 1970).

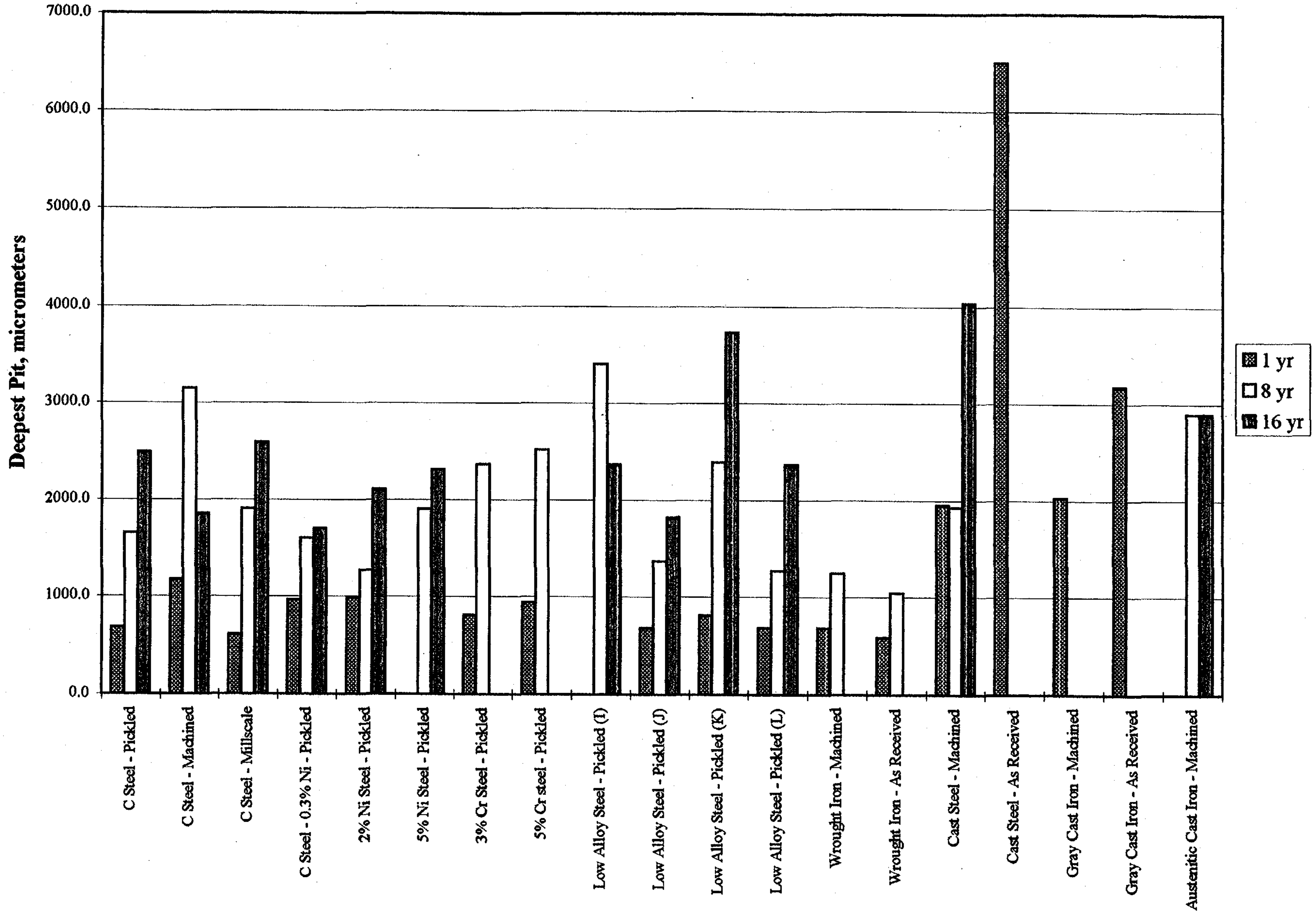

Figure III-5 Maximum pit depth of structural ferrous metals at mean tide elevation - Pacific Ocean, Panama Canal Zone (Southwell Alexander, 1970).

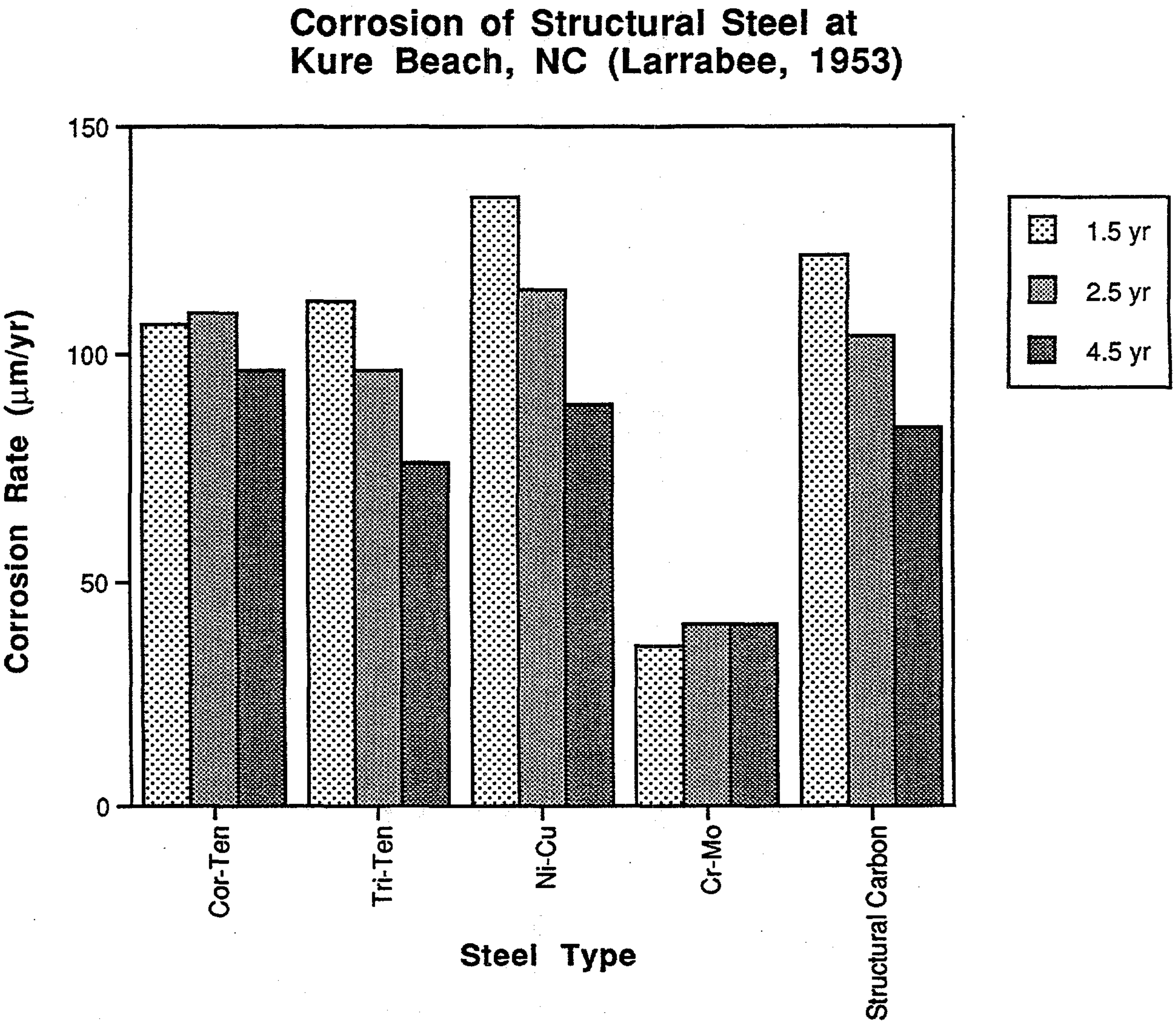

Figure III-6 Corrosion of structural steel at Kure Beach, NC (Larrabee, 1953).

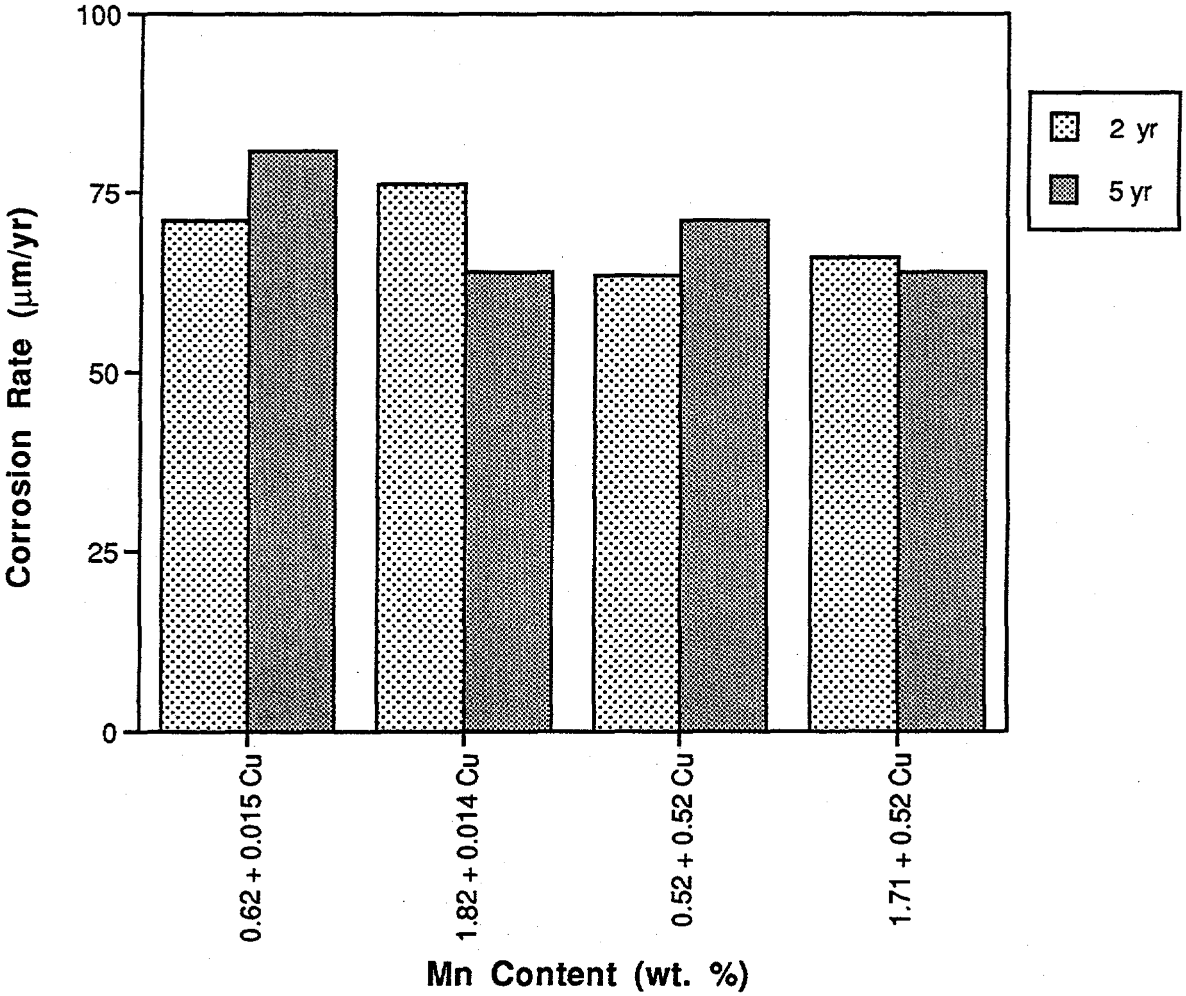

Figure III-7 Aqueous corrosion of manganese steels in sea water at Emsworth, England (Hudson & Stanners, 1955).

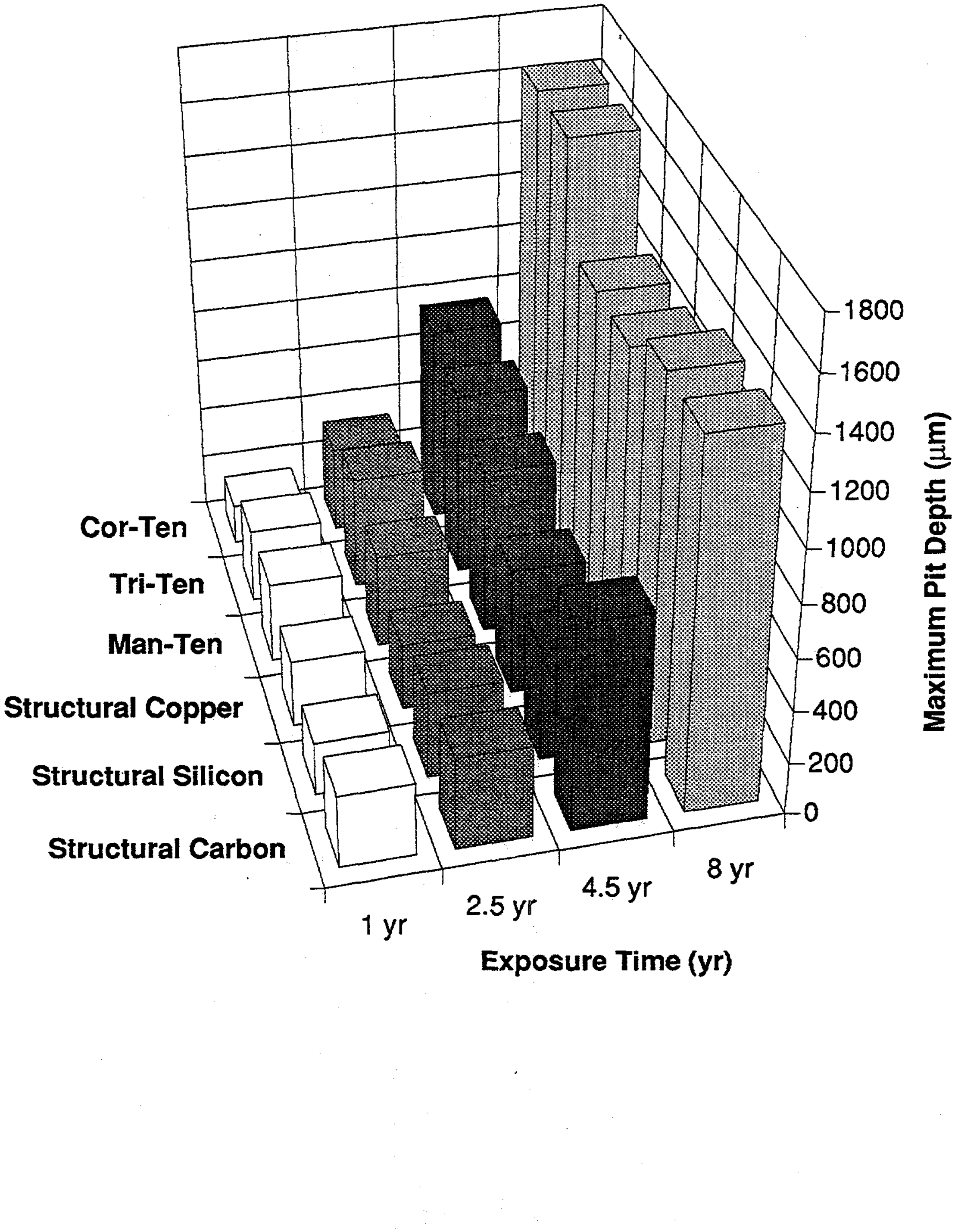

Figure III-9 Pitting of structural steel in the Mississippi River (Larrabee, 1953).

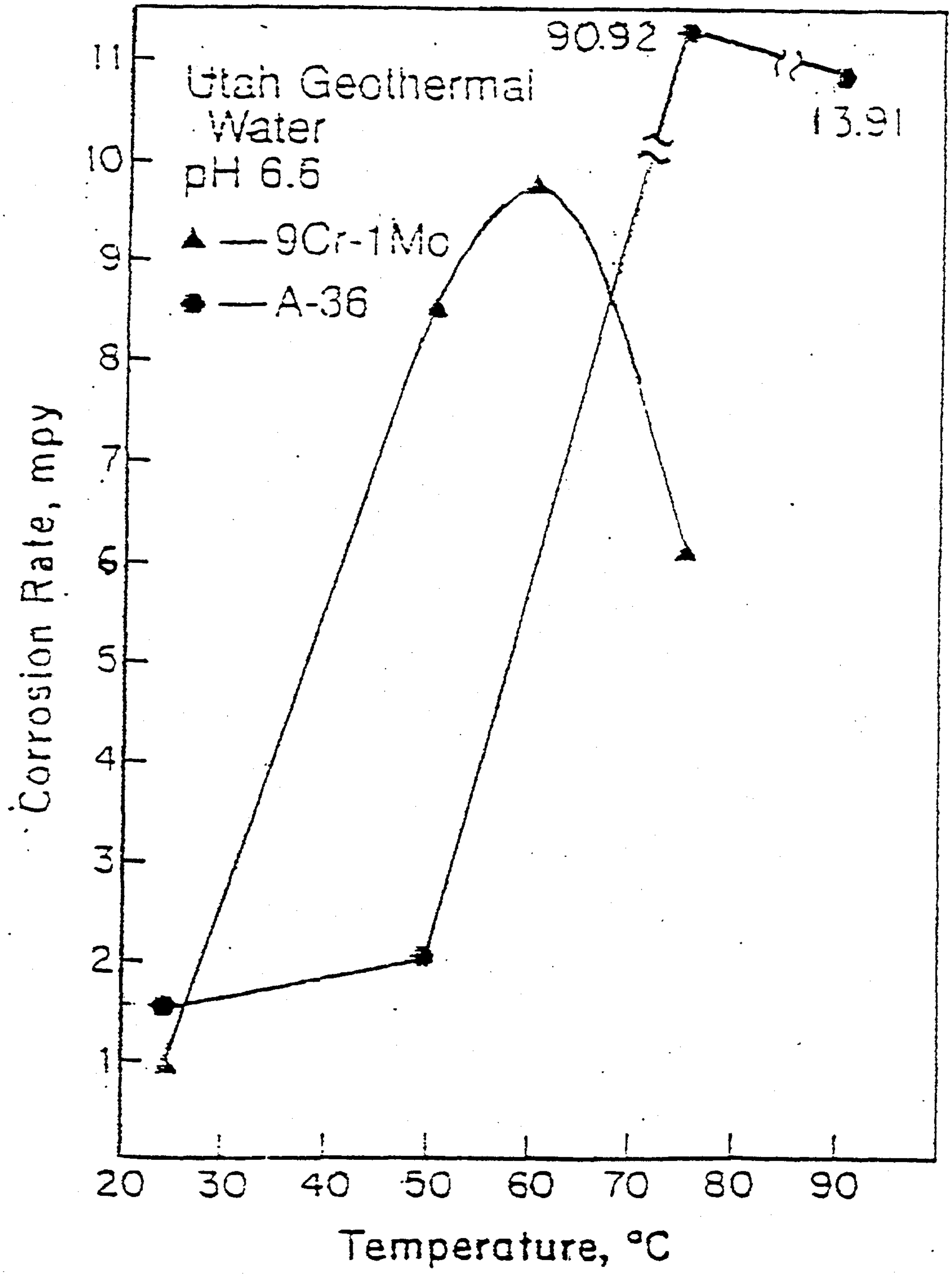

Figure III-10 Corrosion rate of steel alloys exposed to Utah geothermal waters (Hong & Pitt, 1987).

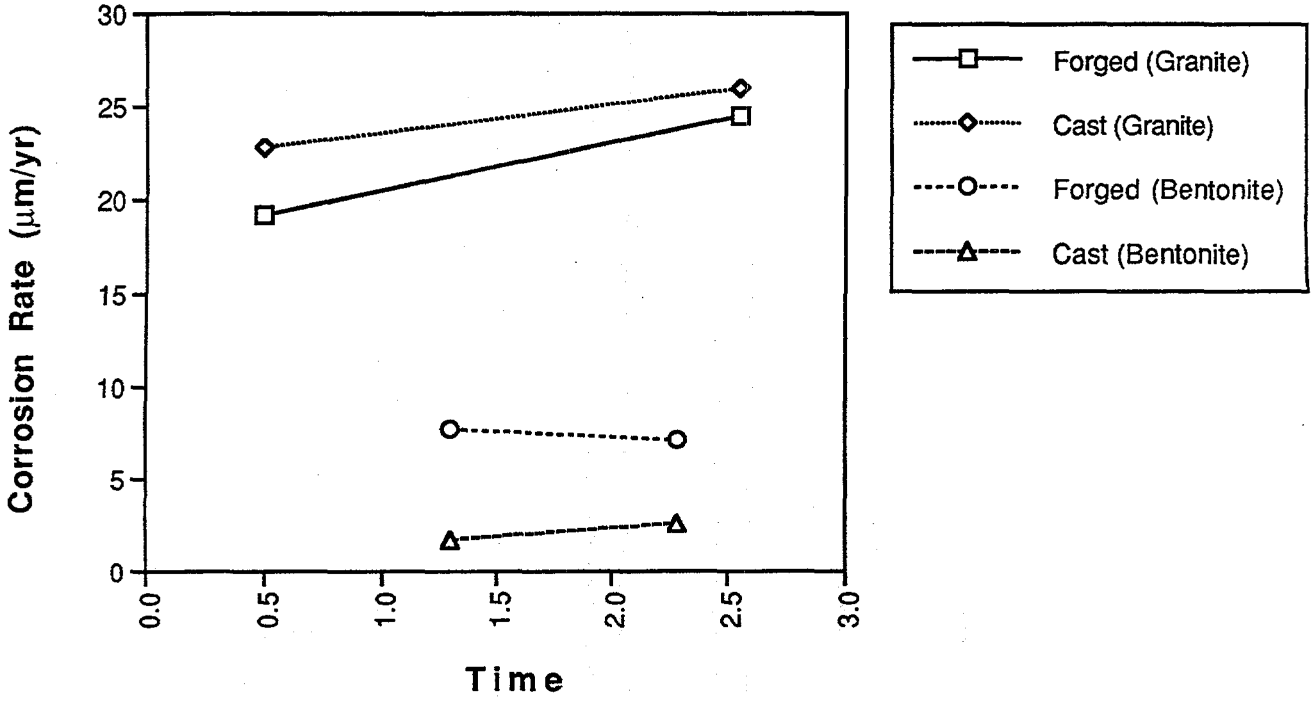

Figure III-11 Corrosion rate for forged and cast mild steels in granite and Bentonite (Marsh & Taylor, 1988).

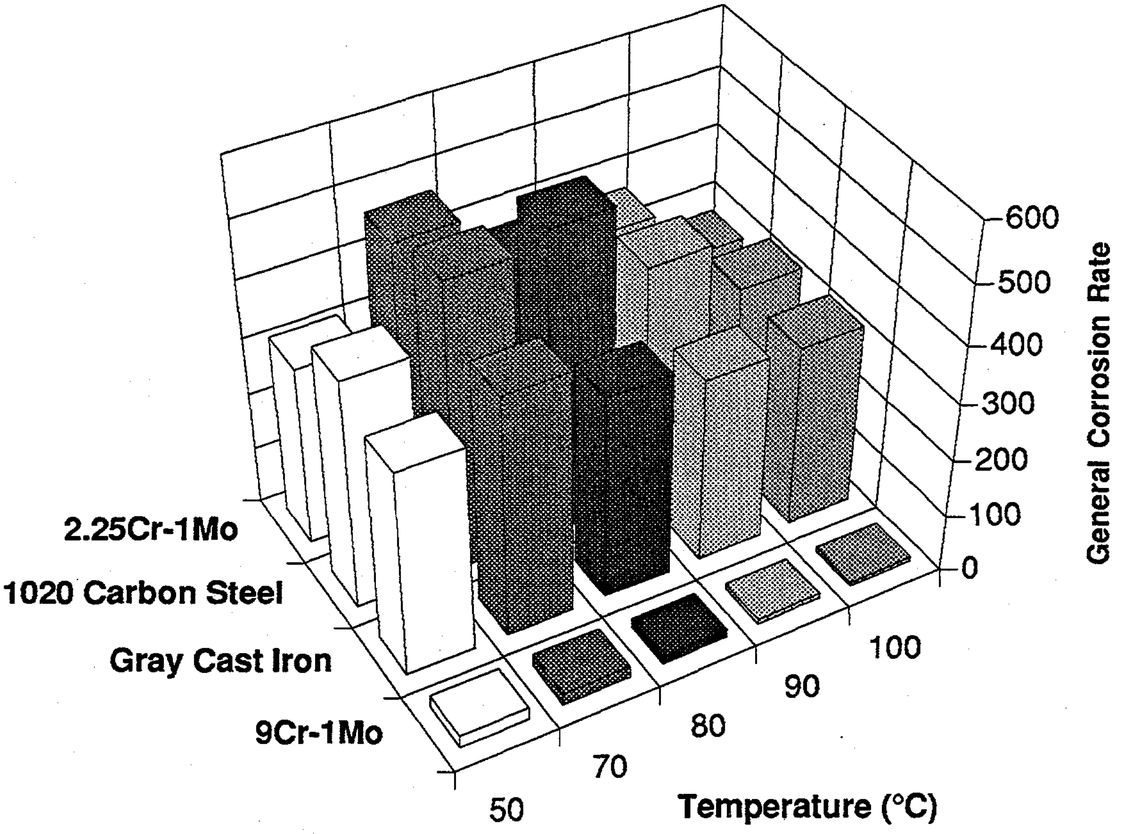

Figure III-12 General corrosion for candidate iron-base materials (McCright and Weiss, 1985).

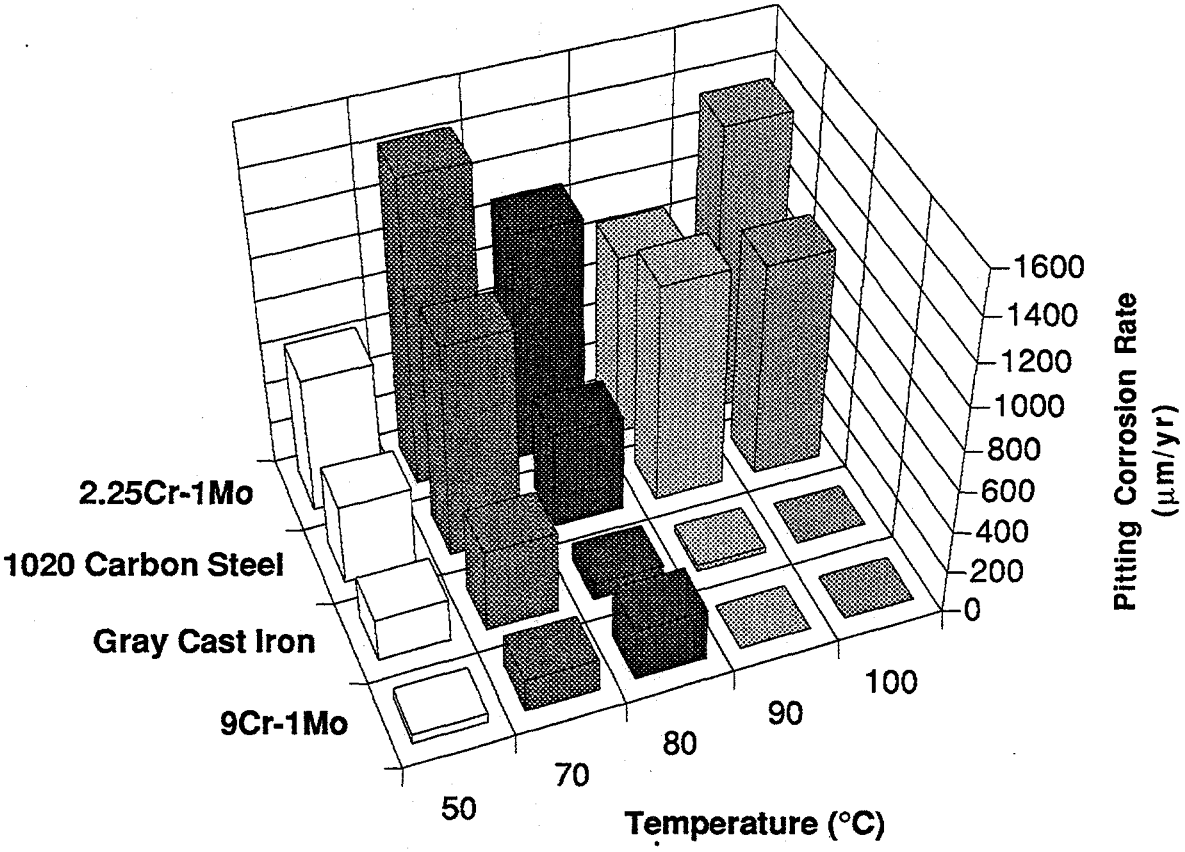

Figure III-13 Pitting corrosion rate for candidate iron-base materials (McCright and Weiss, 1985).

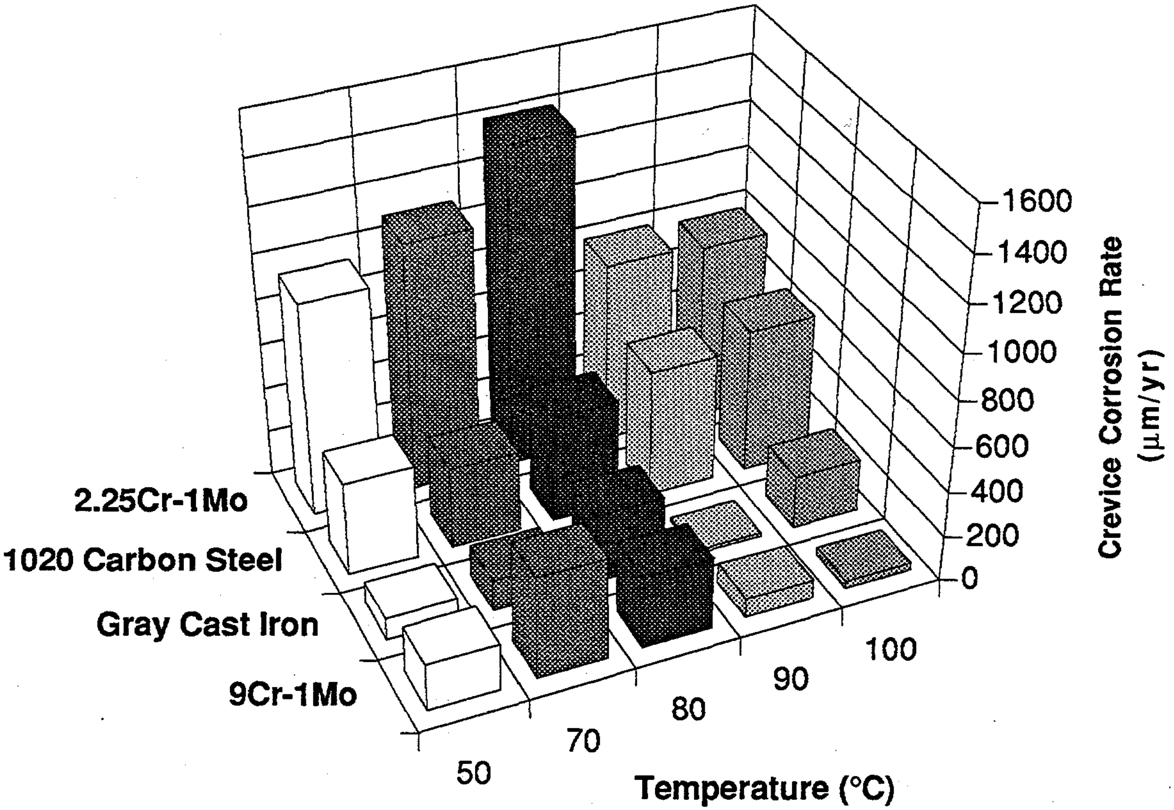

Figure III-14 Crevice corrosion for candidate iron-base materials (McCright and Weiss, 1985).

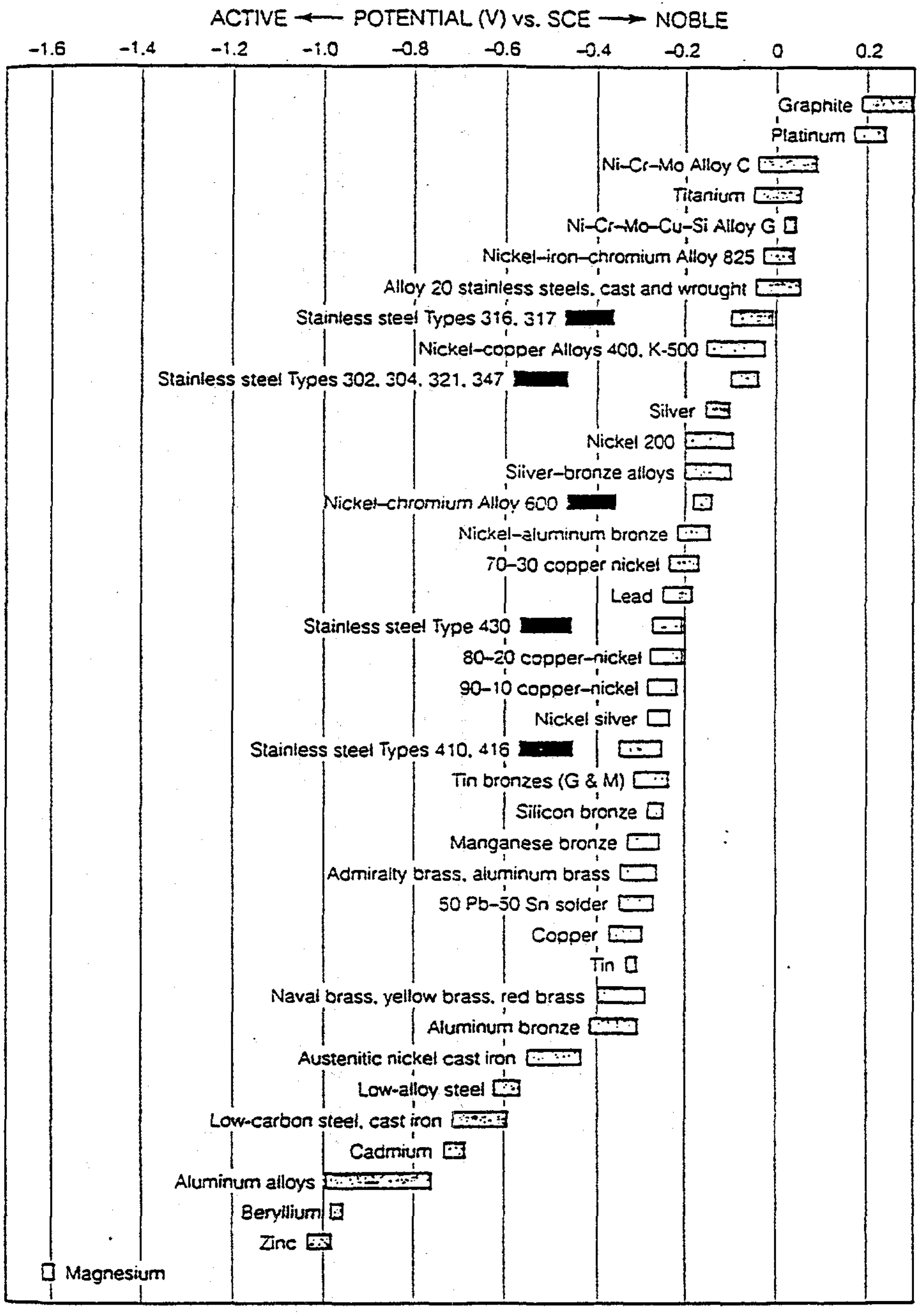

Figure V-1. Galvanic series for several alloys in seawater (*The Corrosion Handbook*, 1948).

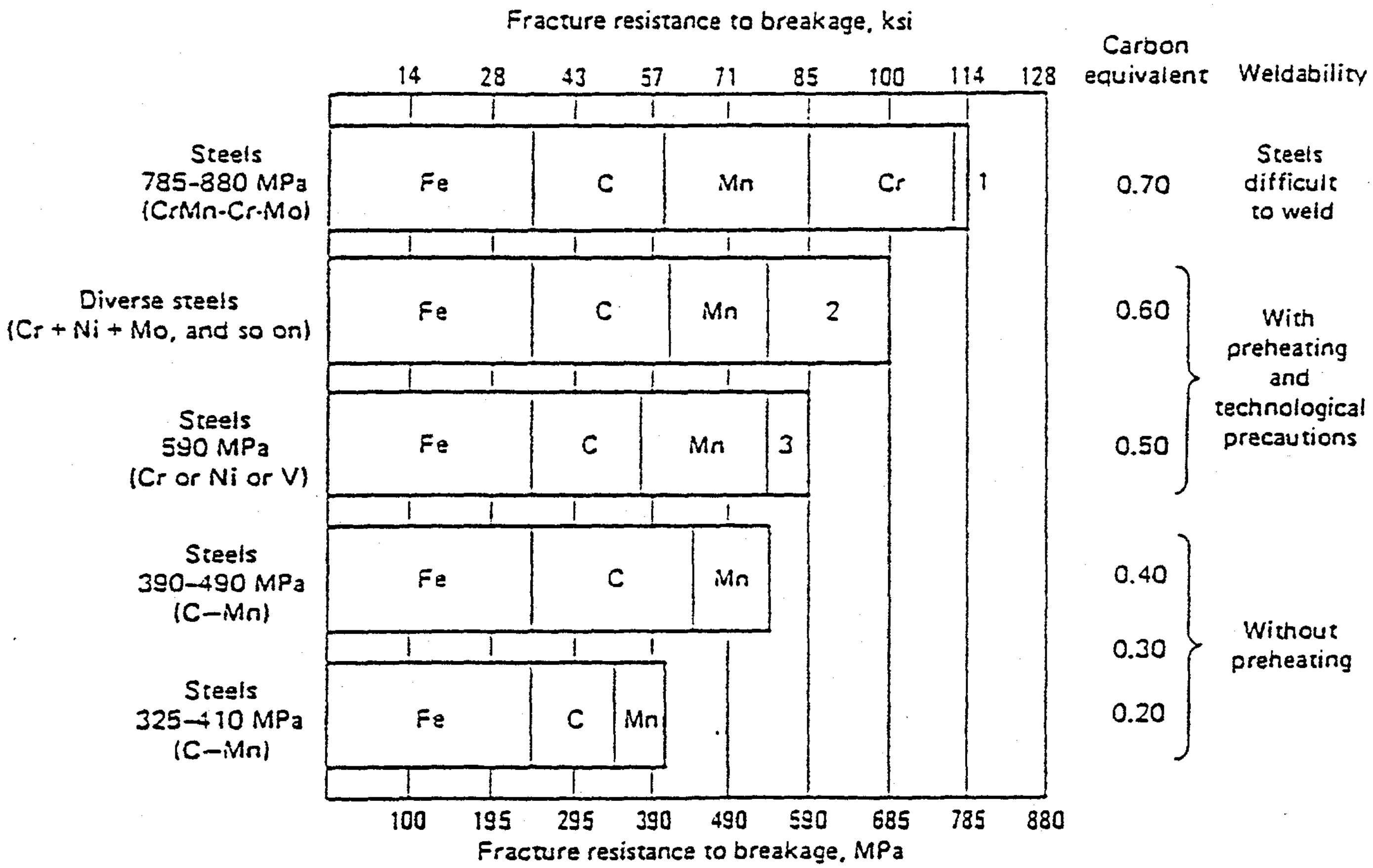

Figure VIII-1 Carbon equivalent and weldability for some families of steel (*Metals Handbook*, 1993).

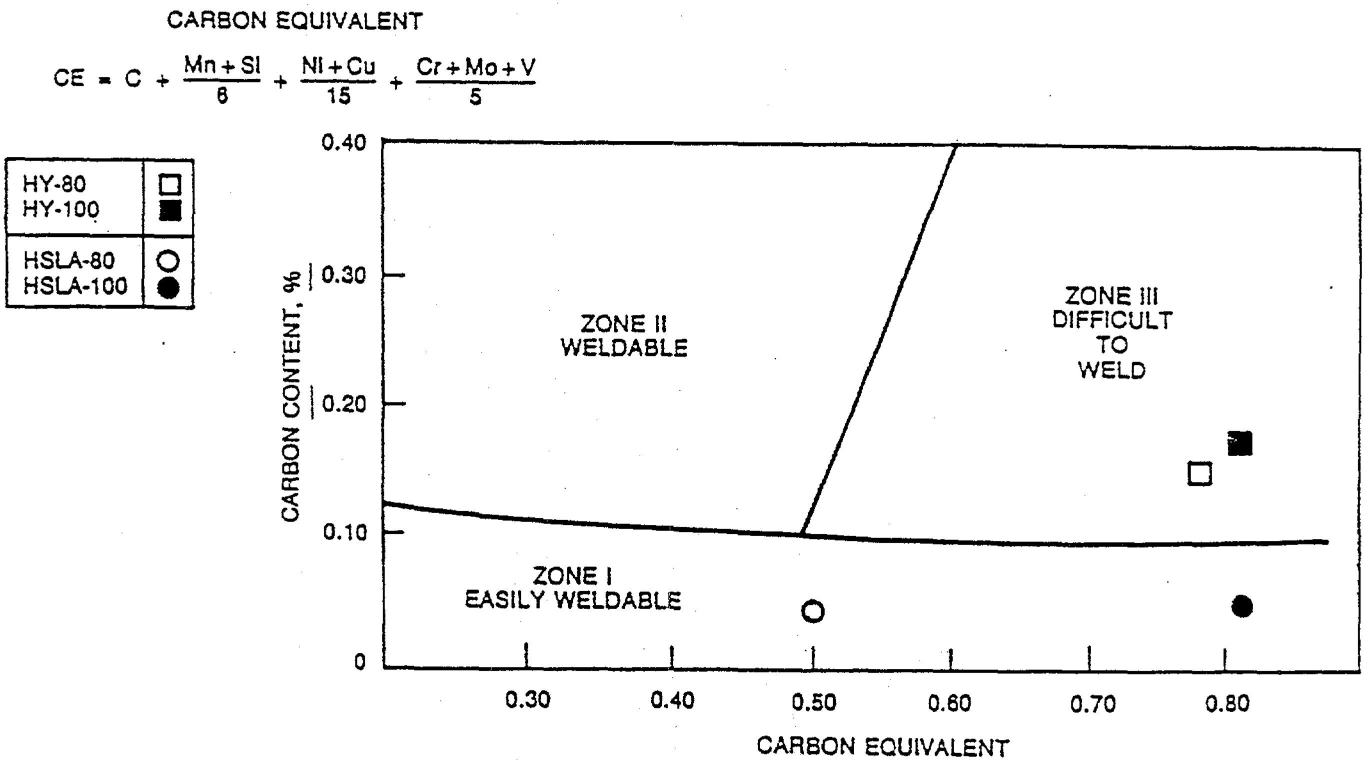

Figure VIII-2. Weldability of iron-base alloys as a funciton of carbon equivalent and composition (Czyryca, 1980).

Table I-A Typical Multiple Purpose Container Dimensions

Corrosion Allowance Outer Barrier	Dimension (cm)
Outer Diameter	175.10
Inner Diameter	155.10
Inner Height	492.90
Outer Height	512.90
Wall Thickness	10.00
Air Gap	0.60

Corrosion Resistant Inner Barrier	Dimension (cm)
Outer Diameter	153.90
Inner Diameter	152.00
Inner Height	491.00
Outer Height	492.90
Wall Thickness	0.95

Table II-A Chemical composition of the Atmofix steels and the comparison steels

(Knotkova-Cermakova, Vlckova, and Honzak, 1982)

Steel	Chemical Composition, %										
	C	Mn	Si	P	S	Cr	Ni	Cu	Nb	V	Al
15217--Atmofix 52A	max	0.30	0.25	0.07	max	0.50	0.30	0.30	. . .	. . .	min
	0.12	0.80	0.70	0.15	0.04	1.20	0.60	0.55	. . .	. . .	0.01
15127--Atmofix 52B	0.10	0.90	0.20	max	max	0.40	0.30	0.30	0.02	0.02	min
	0.17	1.20	0.45	0.04	0.04	0.80	0.60	0.55	0.06	0.06	0.01
15117--Atmofix 37	max	0.20	0.10	max	max	0.30	0.20	0.30	. . .	. . .	. . .
	0.12	0.55	0.35	0.05	0.04	0.60	0.40	0.50	. . .	. . .	. . .
15327--Atmofix 60	max	1.10	0.20	max	max	0.60	0.30	0.30	0.03	. . .	min
	0.16	1.50	0.40	0.05	0.04	1.00	0.50	0.50	0.08	. . .	0.01
11304--Comparison Steel	max	0.40	. . .	0.03	0.03	. . .	. . .	. . .	. . .	. . .	. . .
	0.07	. . .	. . .	. . .	. . .	. . .	. . .	. . .	. . .	. . .	. . .

Table II-B Compositions of Steels %Wt

(McKenzie, 1982)

Steel	C	S	P	Mn	Ni	Cr	Si	Cu	Mo	V
Special Mild Steel	0.19	0.034	0.01	0.59	0.27	0.12	0.031	0.22	0.04	---
Cor-Ten B	0.17	0.021	0.025	1.28	---	0.57	0.28	0.31	---	0.075

Table II-C Compositions of Iron and Steels Subjected to Atmospheric Corrosion

(Briggs, 1968)

Material Tested	Ni	Cu	Mn	Cr	V	Ti	C	Mo	P	S	Si
Cast Steels											
Carbon-Grade A	0.10	0.13	0.61	0.21	0.03	...	0.14	Trace	0.016	0.026	0.41
Ni-Cr-Mo	0.56	0.17	0.80	0.60	0.04	...	0.26	0.15	...	...	0.44
1Ni-1.7Mn	1.08	0.08	1.70	0.08	0.04	...	0.27	...	0.020	0.023	0.42
2Ni	2.26	0.12	0.77	0.19	0.03	...	0.17	Trace	0.017	0.021	0.65
Carbon-Grade B	0.03	0.03	0.65	0.10	0.04	...	0.25	...	0.011	0.021	0.51
1Cu	0.04	0.94	0.87	0.11	0.07	...	0.28	...	...	...	0.42
1.36Mn-0.09V	<0.01	0.15	1.36	0.08	0.09	...	0.37	...	0.031	0.038	0.34
1.42Mn	<0.01	0.13	1.42	0.16	0.04	...	0.37	...	0.027	0.022	0.38
1.5Mn-0.05Ti	<0.01	0.11	1.48	0.04	0.03	0.05	0.33	...	0.016	0.025	0.40
Malleable Iron											
Pearlitic malleable Grade 45010	0.02	0.10	0.87	0.03	...	...	...	nil	0.145	0.086	1.16
Wrought Materials											
AISI 1020, wrought	0.03	0.05	0.79	0.04	...	...	0.18	nil	0.012	0.030	0.01
Corten steel, wrought	0.49	0.46	0.32	1.02	...	...	0.09	0.07	0.120	0.042	0.42

Table II-D Corrosion Performance of Cor-Ten B Steel - Free Exposure - 5 Years (KilCullen & McKenzie, 1978)

Site	Corrosion Loss in First Year of Test, μm	Corrosion Loss in Last Year of Test, μm
Battersea	56.3	27.6
Stratford	69.4	28.8
Teesside	67.7	26.3
Motherwell	55.0	22.3
Sheffield	70.0	35.3
Birmingham	75.0	35.8
Shoreham	52.0	9.1
Rye	52.5	13.8
Leatherhead	45.0	14.5
Brixham	31.4	7.1
Avon Dam	40.9	14.5

Table II-E Corrosion Performance of Cor-Ten B Steel - Sheltered Exposure (Underbridge) - 3 Years

(KilCullen & McKenzie, 1978)

Site	Subject to deicing salt	Loss in First Year of Test, µm	Loss in Last Year of Test, µm
Iden Bridge	No	22.1	17.5
M1 Motorway Bridge	Yes	37.8	25.2
M4 Motorway Bridge	Yes	43.1	25.3

Table II-F Composition of Test Materials exposed to the Atmosphere
(Mannweiler, 1961)

Material	Element %								
	C	Si	Mn	S	P	Cr	Ni	Mo	Cu
ASTM A 47-52, grade 35018									
35018 M.I.-(L)	...	1.12	0.32	0.075	0.155	0.01	0.00	0.00	0.08
35018 M.I.-(E)	...	1.11	0.36	0.070	0.140	0.01	0.00	0.00	0.10
35018 M.I. + Cu	...	1.12	0.36	0.095	0.150	0.01	0.01	0.00	0.65
ASTM A 47-52, grade 32510									
32510 M.I.-(N)	...	1.11	0.57	0.214	0.054	0.02	0.01	0.00	0.14
32510 M.I.-(A)	...	1.60	0.45	0.156	0.037	0.02	0.01	0.00	0.08
ASTM A 220-55T, grade 45010									
45010 P.M.I.-(L)	...	1.14	0.36	0.072	0.153	0.01	0.01	0.00	0.06
45010 P.M.I.-Mn	...	1.16	0.87	0.086	0.145	0.03	0.02	0.00	0.10
ASTM A 220-55T, grade 60003									
60003 P.M.I.-(L)	...	1.12	0.35	0.070	0.155	0.02	0.00	0.00	0.07
60003 P.M.I.-(C)	...	1.31	0.45	0.115	0.054	0.02	0.05	0.00	0.05
ASTM A 220-55T, grade 80002									
80002 P.M.I.-(W)	Material Composition Not Indicated								
ASTM A 339-55, grade 60-45-10									
60-45-10 N.I.	3.57	2.32	0.30	0.018	0.025	0.03	0.08	0.05	0.07
ASTM A 339-55, grade 80-60-03									
80-60-03 N.I.	3.70	2.37	0.28	0.023	0.027	0.01	0.08	0.04	0.05
Mild steel, ASTM 1020	0.18	0.01	0.79	0.030	0.012	0.04	0.03	0.00	0.05
Copper-bearing steel	0.18	0.04	0.63	0.045	0.015	0.03	0.01	0.00	0.26
Cor-Ten steel	0.09	0.42	0.32	0.042	0.120	1.02	0.49	0.07	0.46

Table II-G Chemicals Satisfactorily Handled with High Silicon Cast Iron

(Peacock & Cangi, 1973)

- Aluminum chloride
- Aluminum sulfate
- Barium chloride
- Barium nitrate
- Barium sulfate
- Calcium chlorate
- Calcium chloride
- Calcium hypochloride
- Calcium phosphate
- Ferric nitrate
- Ferric sulfate
- Ferrous chloride
- Ferrous sulfate
- Hydrocarbons
- Hydrochloric acid
- Hydrogen peroxide
- Magnesium chloride
- Magnesium sulfate
- Nitric acid
- Nitrous acid
- Phosphoric acid
- Potassium bisulfate
- Potassium chloride
- Potassium hydroxide
- Potassium nitrate
- Potassium sulfate
- Sodium bicarbonate
- Sodium bisulfate
- Sodium chlorate
- Sodium chloride
- Sodium hydroxide
- Sodium nitrate
- Sodium phosphate
- Sodium sulfate
- Sodium sulfide
- Sodium thiosulfate
- Sulfur
- Sulfuric acid

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961)

		Composition*				Average Reduction of Thickness, µm†		
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Keamy, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
1	R‡	R‡	0.012	R‡	R‡	732§	312	1321§
2	R	R	0.040	R	R	224	201	363
3	R	R	0.100	R	R	201	198	305
4	R	R	0.240	R	R	155	163	284
5	R	R	0.450	R	R	135	152	264
6	R	R	0.008	1.0	R	155	132	244
7	R	R	0.050	1.0	R	140	130	228
8	R	R	0.110	1.0	R	122	124	208
9	R	R	0.200	1.0	R	112	117	203
10	R	R	0.470	1.0	R	99	104	183
11	R	R	0.012	R	0.61	1059§	419	401§
12	R	R	0.060	R	0.69	224	185	254
13	R	R	0.100	R	0.64	152	163	249
14	R	R	0.220	R	0.63	117	145	229
15	R	R	0.480	R	0.61	99	130	208
16	R	R	0.012	1.0	0.60	137	127	185
17	R	R	0.060	1.0	0.63	102	112	173
18	R	R	0.120	1.0	0.60	89	107	165
19	R	R	0.220	1.0	0.59	74	91	163
20	R	R	0.460	1.0	0.60	69	81	150
21	R	R	0.016	R	1.30	419	287	478
22	R	R	0.050	R	1.30	218	168	231
23	R	R	0.120	R	1.30	109	127	208
24	R	R	0.220	R	1.30	89	114	211
25	R	R	0.460	R	1.20	74	104	198
26	R	R	0.016	1.0	1.30	102	104	147
27	R	R	0.060	1.1	1.30	84	86	140
28	R	R	0.120	1.0	1.20	69	81	130
29	R	R	0.210	1.0	1.20	61	76	127
30	R	R	0.440	1.1	1.20	56	74	119
31	R	0.22	0.012	R	R	373	257	546§
32	R	0.23	0.060	R	R	198	180	279
33	R	0.18	0.100	R	R	170	175	264
34	R	0.20	0.220	R	R	152	155	251
35	R	0.18	0.480	R	R	127	137	226
36	R	0.23	0.020	1.0	R	132	124	224
37	R	0.22	0.050	1.0	R	114	117	203
38	R	0.27	0.120	1.0	R	109	112	185
39	R	0.24	0.220	1.0	R	107	109	183
40	R	0.18	0.460	1.1	R	97	99	173

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961) -- (Continued)

	Composition*					Average Reduction of Thickness, µm†		
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Kearny, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
41	R	0.31	0.016	R	0.66	485	351	381§
42	R	0.27	0.060	R	0.69	196	188	221
43	R	0.26	0.110	R	0.69	135	175	216
44	R	0.16	0.200	R	0.68	109	157	216
45	R	0.15	0.470	R	0.65	91	137	203
46	R	0.25	0.012	1.0	0.68	112	119	163
47	R	0.24	0.060	1.0	0.70	84	102	150
48	R	0.15	0.100	1.0	0.66	84	104	152
49	R	0.15	0.220	1.1	0.67	71	97	145
50	R	0.21	0.470	1.0	0.66	66	84	130
51	R	0.27	0.016	R	1.30	353	246	345
52	R	0.29	0.050	R	1.20	211	160	196
53	R	0.24	0.100	R	1.20	145	137	193
54	R	0.16	0.200	R	1.20	89	119	196
55	R	0.15	0.480	R	1.20	74	107	185
56	R	0.24	0.012	1.0	1.30	86	89	130
57	R	0.25	0.050	1.0	1.20	79	81	124
58	R	0.22	0.100	1.0	1.30	64	74	117
59	R	0.24	0.220	1.0	1.20	53	69	109
60	R	0.19	0.460	1.0	1.20	48	64	112
61	R	0.61	0.016	R	R	361	213	333
62	R	0.56	0.060	R	R	170	163	213
63	R	0.47	0.110	R	R	145	163	218
64	R	0.47	0.220	R	R	130	152	208
65	R	0.47	0.510	R	R	117	130	185
66	R	0.57	0.016	1.0	R	109	109	152
67	R	0.47	0.044	1.0	R	104	104	165
68	R	0.47	0.090	1.0	R	107	107	155
69	R	0.52	0.240	1.0	R	102	97	142
70	R	0.52	0.480	1.0	R	94	89	124
71	R	0.59	0.016	R	0.64	312	216	201
72	R	0.51	0.060	R	0.68	152	150	180
73	R	0.53	0.110	R	0.68	104	127	160
74	R	0.61	0.200	R	0.76	79	102	140
75	R	0.52	0.460	R	0.69	76	107	145
76	R	0.49	0.016	1.0	0.71	91	94	127
77	R	0.50	0.050	1.0	0.67	79	86	119
78	R	0.58	0.120	1.1	0.62	71	74	107
79	R	0.62	0.220	1.0	0.69	64	71	104
80	R	0.50	0.480	1.0	0.67	58	69	107

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961) -- (Continued)

			Composition*				Average Reduction of Thickness, µm†	
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Kearny, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
81	R	0.56	0.016	R	1.30	236	160	206
82	R	0.54	0.060	R	1.30	150	104	147
83	R	0.54	0.110	R	1.30	86	86	135
84	R	0.58	0.200	R	1.30	66	76	124
85	R	0.50	0.440	R	1.20	61	81	137
86	R	0.51	0.020	0.9	1.30	79	74	109
87	R	0.48	0.050	1.0	1.20	69	61	102
88	R	0.56	0.110	1.0	1.30	56	53	89
89	R	0.46	0.220	1.0	1.30	48	53	94
90	R	0.44	0.440	0.9	1.20	48	53	102
91	0.06	R	0.020	R	R	198	175	358
92	0.06	R	0.060	R	R	150	147	257
93	0.06	R	0.110	R	R	135	147	244
94	0.06	R	0.210	R	R	124	130	231
95	0.07	R	0.480	R	R	99	107	193
96	0.06	R	0.020	1.1	R	104	107	206
97	0.06	R	0.060	1.0	R	102	102	191
98	0.06	R	0.110	1.1	R	97	102	183
99	0.06	R	0.210	1.0	R	89	91	173
100	0.06	R	0.480	1.1	R	79	79	152
101	0.06	R	0.020	R	0.66	282	216	195
102	0.06	R	0.050	R	0.66	191	160	131
103	0.06	R	0.120	R	0.63	117	132	211
104	0.06	R	0.220	R	0.63	91	112	193
105	0.06	R	0.450	R	0.64	79	99	173
106	0.06	R	0.016	1.0	0.60	102	102	165
107	0.06	R	0.060	1.1	0.62	86	89	157
108	0.06	R	0.100	1.1	0.62	79	91	150
109	0.06	R	0.240	1.0	0.66	660	76	132
110	0.06	R	0.450	1.0	0.65	64	74	127
111	0.06	R	0.020	R	1.30	221	173	229
112	0.06	R	0.060	R	1.30	140	124	188
113	0.05	R	0.110	R	1.30	102	109	178
114	0.06	R	0.230	R	1.30	66	89	157
115	0.06	R	0.460	R	1.30	66	86	155
116	0.06	R	0.020	1.0	1.20	89	86	127
117	0.06	R	0.060	1.0	1.30	74	74	127
118	0.06	R	0.100	1.0	1.30	61	69	117
119	0.07	R	0.200	1.0	1.20	51	64	107
120	0.06	R	0.470	1.0	1.20	51	56	104

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961) -- (Continued)

	Composition*					Average Reduction of Thickness, µm†		
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Kearny, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
121	0.06	0.25	0.016	R	R	241	175	295
122	0.06	0.25	0.050	R	R	163	140	236
123	0.06	0.18	0.100	R	R	145	135	234
124	0.06	0.25	0.210	R	R	124	122	208
125	0.06	0.18	0.470	R	R	112	112	198
126	0.06	0.20	0.020	1.0	R	107	104	193
127	0.06	0.22	0.050	1.0	R	104	102	180
128	0.06	0.20	0.100	1.0	R	102	97	173
129	0.06	0.19	0.200	1.0	R	99	91	160
130	0.06	0.18	0.460	1.1	R	81	81	150
131	0.06	0.27	0.012	R	0.58	305	234	267
132	0.06	0.18	0.044	R	0.61	178	163	203
133	0.06	0.26	0.100	R	0.61	104	130	173
134	0.06	0.19	0.170	R	0.66	94	124	180
135	0.06	0.22	0.450	R	0.65	76	102	157
136	0.07	0.20	0.012	1.0	0.63	97	99	140
137	0.06	0.17	0.042	1.1	0.61	84	94	137
138	0.06	0.23	0.080	1.0	0.66	71	81	119
139	0.06	0.15	0.220	1.1	0.53	74	86	135
140	0.05	0.17	0.450	1.0	0.65	64	76	117
141	0.06	0.30	0.008	R	1.30	241	175	191
142	0.05	0.24	0.044	R	1.30	147	119	155
143	0.06	0.25	0.120	R	1.30	74	86	137
144	0.06	0.22	0.200	R	1.20	69	89	142
145	0.05	0.17	0.430	R	1.30	64	84	142
146	0.07	0.29	0.012	1.0	1.20	77	71	109
147	0.06	0.24	0.044	1.1	1.30	71	66	109
148	0.06	0.23	0.090	1.0	1.30	58	61	102
149	0.05	0.17	0.210	1.0	1.20	56	61	107
150	0.06	0.15	0.440	1.1	1.30	46	56	99
151	0.06	0.53	0.012	R	R	241	173	241
152	0.06	0.50	0.044	R	R	150	140	198
153	0.06	0.53	0.080	R	R	114	124	178
154	0.06	0.51	0.220	R	R	94	107	157
155	0.06	0.42	0.450	R	R	104	109	168
156	0.06	0.47	0.012	1.0	R	97	91	160
157	0.07	0.52	0.050	1.0	R	89	84	140
158	0.06	0.54	0.120	1.0	R	86	84	130
159	0.06	0.56	0.230	1.0	R	81	81	122
160	0.06	0.51	0.450	1.0	R	81	79	117

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961) -- (Continued)

	Composition*					Average Reduction of Thickness, µm†		
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Keamy, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
161	0.06	0.53	0.016	R	0.58	264	180	193
162	0.07	0.49	0.009	R	0.59	244	168	180
163	0.06	0.57	0.100	R	0.66	86	97	127
164	0.06	0.55	0.220	R	0.67	66	86	119
165	0.06	0.50	0.450	R	0.65	64	79	117
166	0.07	0.53	0.016	1.1	0.52	84	74	109
167	0.06	0.47	0.050	1.0	0.64	76	71	114
168	0.06	0.48	0.090	1.0	0.70	69	64	104
169	0.06	0.57	0.230	1.0	0.67	58	56	89
170	0.06	0.47	0.440	1.1	0.67	56	56	91
171	0.06	0.64	0.012	R	1.20	193	109	130
172	0.06	0.53	0.060	R	1.30	94	71	104
173	0.06	0.52	0.100	R	1.20	81	69	107
174	0.06	0.56	0.200	R	1.30	58	58	102
175	0.06	0.46	0.450	R	1.30	53	64	114
176	0.06	0.46	0.008	1.0	1.20	71	53	97
177	0.06	0.52	0.044	1.0	1.30	66	53	89
178	0.07	0.54	0.100	1.0	1.00	48	43	79
179	0.06	0.48	0.210	1.0	1.20	48	46	84
180	0.06	0.49	0.450	1.0	1.30	43	43	84
181	0.10	R	0.012	R	R	191	160	333
182	0.10	R	0.060	R	R	132	119	239
183	0.09	R	0.110	R	R	119	117	218
184	0.10	R	0.200	R	R	112	109	213
185	0.10	R	0.490	R	R	91	91	180
186	0.09	R	0.019	1.0	R	102	89	183
187	0.09	R	0.060	1.0	R	91	86	170
188	0.09	R	0.090	1.0	R	86	84	168
189	0.09	R	0.210	1.1	R	79	79	152
190	0.09	R	0.430	0.9	R	71	69	142
191	0.08	R	0.020	R	0.70	249	188	239
192	0.10	R	0.044	R	0.59	168	135	208
193	0.09	R	0.110	R	0.63	91	114	178
194	0.09	R	0.220	R	0.64	81	99	173
195	0.09	R	0.450	R	0.63	74	91	152
196	0.09	R	0.012	1.1	0.64	97	94	152
197	0.12	R	0.050	1.0	0.58	74	79	137
198	0.10	R	0.110	1.0	0.65	66	71	127
199	0.10	R	0.210	1.0	0.64	58	71	122
200	0.09	R	0.440	1.0	0.62	56	66	117

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961) -- (Continued)

			Composition*				Average Reduction of Thickness, µm†	
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Kearny, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
201	0.08	R	0.016	R	1.30	203	155	201
202	0.10	R	0.040	R	1.20	137	109	165
203	0.11	R	0.100	R	1.10	89	91	150
204	0.08	R	0.200	R	1.20	64	79	147
205	0.08	R	0.460	R	1.30	58	69	137
206	0.11	R	0.016	1.0	1.20	84	74	117
207	0.09	R	0.035	1.0	1.30	66	64	107
208	0.08	R	0.110	1.0	1.20	58	58	107
209	0.09	R	0.180	1.0	1.30	48	53	97
210	0.08	R	0.410	1.0	1.20	48	53	99
211	0.09	0.26	0.024	R	R	191	157	262
212	0.08	0.24	0.050	R	R	145	132	229
213	0.09	0.18	0.080	R	R	124	117	213
214	0.09	0.29	0.220	R	R	112	112	185
215	0.10	0.18	0.480	R	R	104	99	175
216	0.10	0.31	0.024	0.9	R	99	86	157
217	0.09	0.29	0.060	1.0	R	91	91	160
218	0.08	0.23	0.100	1.0	R	94	91	160
219	0.11	0.16	0.200	1.0	R	84	81	145
220	0.08	0.24	0.440	1.1	R	74	76	135
221	0.09	0.23	0.016	R	0.62	224	185	216
222	0.08	0.27	0.050	R	0.63	119	130	170
223	0.08	0.26	0.090	R	0.63	89	114	160
224	0.09	0.28	0.200	R	0.68	71	91	142
225	0.09	0.26	0.410	R	0.65	71	94	140
226	0.08	0.26	0.012	1.1	0.62	86	89	130
227	0.08	0.23	0.040	1.0	0.70	76	79	127
228	0.09	0.27	0.100	1.0	0.69	66	71	112
229	0.10	0.17	0.210	1.0	0.61	58	71	117
230	0.11	0.17	0.420	1.0	0.61	56	61	102
231	0.11	0.32	0.014	R	1.20	175	127	152
232	0.10	0.28	0.050	R	1.20	132	102	140
233	0.08	0.23	0.090	R	1.30	74	84	135
234	0.09	0.27	0.220	R	1.30	58	69	119
235	0.09	0.27	0.480	R	1.30	53	66	117
236	0.10	0.21	0.020	1.0	1.20	71	58	107
237	0.08	0.20	0.040	1.0	1.20	66	56	104
238	0.09	0.23	0.100	1.0	1.30	56	53	97
239	0.10	0.18	0.210	1.0	1.20	48	51	97
240	0.08	0.25	0.450	1.0	1.30	46	46	97

Table II-H Atmospheric Corrosion Data for 270 Materials Tested at Three Sites for a Period of 15.5 Years

(Larrabee & Coburn, 1961) -- (Continued)

	Composition*					Average Reduction of Thickness, µm†		
Material Identification	%P	%Si	%Cu	%Ni	%Cr	Kearny, N.J. INDUSTRIAL	South Bend, PA SEMI-RURAL	Kure Beach, N.C. MARINE
241	0.10	0.54	0.012	R	R	160	132	198
242	0.10	0.54	0.050	R	R	114	112	168
243	0.11	0.48	0.100	R	R	102	102	160
244	0.10	0.53	0.230	R	R	94	94	147
245	0.10	0.56	0.490	R	R	84	89	127
246	0.10	0.46	0.008	1.0	R	91	81	145
247	0.11	0.46	0.050	1.0	R	84	74	135
248	0.11	0.53	0.090	1.0	R	81	71	124
249	0.10	0.47	0.180	1.0	R	79	71	117
250	0.08	0.60	0.450	1.0	R	74	64	99
251	0.11	0.55	0.012	R	0.59	191	132	152
252	0.10	0.55	0.050	R	0.62	114	94	119
253	0.08	0.53	0.090	R	0.68	76	81	119
254	0.11	0.60	0.210	R	0.62	58	61	99
255	0.08	0.47	0.440	R	0.57	66	81	119
256	0.10	0.51	0.020	1.0	0.65	76	64	109
257	0.09	0.42	0.060	1.1	0.66	66	61	109
258	0.09	0.50	0.100	1.1	0.69	58	53	99
259	0.10	0.50	0.210	1.1	0.66	53	51	94
260	0.09	0.62	0.450	1.0	0.67	51	48	89
261	0.11	0.56	0.016	R	1.20	157	91	117
262	0.11	0.54	0.050	R	1.20	97	66	99
263	0.08	0.58	0.080	R	1.30	66	58	97
264	0.10	0.62	0.210	R	1.20	48	43	84
265	0.11	0.60	0.440	R	1.20	43	43	89
266	0.12	0.50	0.008	1.0	1.20	66	48	99
267	0.11	0.58	0.050	1.0	1.20	61	43	84
268	0.08	0.57	0.100	1.0	1.20	51	41	79
269	0.09	0.43	0.240	1.0	1.30	48	43	89
270	0.09	0.59	0.460	1.0	1.30	41	33	81

* In all steels C was less than 0.1, Mn was 0.25 to 0.40, and S was less than 0.020.

† Average reduction of thickness is calculated from loss of weight.

‡ R = 0.010 or less for P, 0.1 or less for Si, 0.1 or less for Cr, and 0.05 or less for Ni.

§ Estimated from time-corrosion curve of earlier losses because this specimen had corroded so badly it had fallen from the rack and was lost.

Table II-I Composition of Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; *1945, 1948, 1952, 1960*)

Material ID Number	Material Classification	Composition, percent C	Mn	P	S	Si	Cu	Ni	Cr	Other
25	High purity iron plus copper	0.049	0.110	0.002	0.040	0.003	0.004	0.003	0.050	----
9		0.022	0.035	0.006	0.020	0.002	0.007	0.006	0.001	----
44		0.016	0.016	0.005	0.030	0.002	0.014	0.050	----	----
45		0.020	0.020	0.006	0.030	0.003	0.020	0.050	----	----
63		0.020	0.023	0.005	0.030	0.002	0.053	0.050	----	----
54		0.020	0.070	0.003	0.030	0.010	0.100	0.180	----	----
55		0.010	0.060	0.003	0.030	0.010	0.270	0.130	----	----
10	Low phosphorus steel, plus copper	0.410	0.220	0.010	0.040	0.002	0.020	0.050	0.010	----
31		0.090	0.350	0.009	0.020	0.050	0.041	0.050	----	----
5		0.043	0.390	0.007	0.020	0.007	0.050	0.010	0.070	----
57		0.060	0.260	0.007	0.030	0.010	0.190	0.060	----	0.01 Mo
6		0.046	0.400	0.007	0.020	0.005	0.210	----	0.070	----
41		0.080	0.340	0.008	0.020	0.010	0.220	0.050	----	----
43		0.060	0.320	0.010	0.020	0.010	0.240	0.050	----	----
64		0.026	0.240	0.009	0.040	0.004	0.265	0.050	----	----
40		0.030	0.170	0.006	0.030	0.010	0.270	0.050	----	----
58		0.050	0.310	0.008	0.020	0.010	0.270	0.050	----	----
42		0.080	0.420	0.010	0.020	0.010	0.300	0.050	----	----
56		0.040	0.310	0.004	0.020	0.010	0.300	0.050	----	0.01 Mo
7		0.038	0.400	0.008	0.020	0.004	0.500	0.004	0.070	----
8		0.040	0.390	0.007	0.020	0.005	1.030	0.004	0.060	----
1	High phosphorus steel, plus copper	0.085	0.430	0.103	0.050	0.005	0.010	0.002	0.020	----
26		0.021	0.050	0.105	0.030	0.180	0.010	0.002	----	----
11		0.040	0.490	0.074	0.040	0.003	0.050	0.010	0.100	0.003 Mo
2		0.095	0.410	0.106	0.060	0.007	0.200	0.003	0.030	----
12		0.026	0.470	0.066	0.060	0.003	0.270	0.030	0.030	0.003 Mo
16		0.090	0.430	0.058	0.030	0.005	0.360	0.240	0.060	----
3		0.095	0.410	0.104	0.050	0.007	0.510	0.002	0.020	----
4		0.095	0.400	0.107	0.060	0.005	0.940	0.001	0.020	----
18	High manganese and silicon steels, plus copper	0.330	0.780	0.043	0.040	0.310	0.030	0.030	0.060	----
19		0.160	1.320	0.018	0.030	0.240	0.030	0.030	0.120	0.093 V
17		0.250	1.350	0.022	0.030	0.180	0.280	0.010	0.030	----
66		0.170	0.670	0.012	0.030	0.230	0.290	0.050	0.140	----
53		0.100	0.690	0.072	0.040	0.260	0.430	0.060	----	----
67	Copper steel, plus chromium and silicon	0.070	1.520	0.017	0.020	0.750	0.100	0.080	0.400	----
68		0.220	1.130	0.060	0.020	0.650	0.120	0.040	0.500	----
154		0.170	0.790	0.017	----	0.890	0.220	0.100	0.600	----
97		0.130	0.660	0.029	0.020	0.790	0.220	0.110	0.560	----
13		0.072	0.270	0.140	0.020	0.830	0.460	0.030	1.190	----

Table II-I Composition of Iron and Steel Specimens Exposed to Industrial and Marine Environments of Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) - (Continued)

Material ID Number	Material Classification	Composition, percent								
		C	Mn	P	S	Si	Cu	Ni	Cr	Other
118	Copper steel, plus molybdenum	0.040	0.150	0.009	0.030	0.010	0.420	0.050	----	0.07 Mo
120		0.040	0.120	0.010	0.020	0.010	0.470	0.050	----	0.07 Mo
38		0.210	1.040	0.017	0.030	0.230	0.620	0.036	0.025	0.35 Mo, 0.006 V
121		0.060	0.096	0.069	0.030	0.010	0.540	0.140	----	0.13 Mo
48		0.170	0.890	0.075	0.030	0.050	0.470	0.160	----	0.28 Mo
32	Nickel steel	0.110	0.690	0.011	0.020	0.140	0.038	0.200	----	----
27		0.160	0.560	0.013	0.030	0.230	0.120	0.540	0.050	----
33		0.140	0.370	0.008	0.020	0.070	0.030	0.550	----	----
34		0.140	0.360	0.008	0.020	0.370	0.028	1.270	----	----
28		0.140	0.470	0.014	0.020	0.230	0.030	1.470	0.050	----
147		0.060	0.140	0.013	0.030	0.140	0.030	1.960	----	----
46		0.160	0.570	0.015	0.020	0.020	0.240	2.200	----	----
35		0.140	0.280	0.008	0.020	0.080	0.030	2.550	----	----
20		0.190	0.530	0.016	0.020	0.009	0.070	3.230	----	----
62		0.170	0.580	0.007	0.010	0.260	0.090	4.980	----	----
36		0.130	0.230	0.007	0.010	0.070	0.030	4.990	0.050	----
30	Nickel steel plus chromium	0.130	0.450	0.017	0.030	0.230	0.040	1.180	0.650	0.01 Mo
29	Nickel steel plus molybdenum	0.160	0.530	0.013	0.010	0.250	0.030	1.840	0.090	0.24 Mo
51	Nickel steel plus chromium and molybdenum	0.100	0.590	0.013	0.010	0.490	0.090	1.020	1.010	0.21 Mo
52		0.080	0.570	0.015	0.010	0.330	0.190	1.340	0.740	0.25 Mo
37	Nickel copper steel	0.090	0.660	0.106	0.020	0.044	1.120	0.640	----	----
60		0.120	0.570	0.010	0.020	0.170	1.050	1.000	----	----
148		0.050	0.170	0.014	0.030	0.130	0.890	1.050	----	----
50		0.090	0.480	0.055	0.030	1.000	1.060	1.140	----	----
39		0.110	0.430	0.012	0.020	0.180	1.090	1.520	----	----
149		0.050	0.320	0.043	0.070	0.200	0.860	1.580	----	----
150		0.050	0.130	0.012	0.030	0.130	0.930	1.770	----	----
153		0.060	0.270	0.013	0.010	0.200	1.830	2.780	----	----
152		0.060	0.140	0.015	0.030	0.410	1.300	2.820	----	----
151		0.080	0.200	0.015	0.010	0.210	1.140	3.120	----	----
49	Nickel copper steel plus chromium	0.090	0.330	0.066	0.030	0.010	0.220	0.400	0.190	----
16-1		0.044	0.200	0.010	0.020	0.003	0.540	0.490	0.460	----
59		0.110	0.650	0.086	0.020	0.130	0.570	0.290	0.660	----
47		0.110	0.750	0.020	0.040	0.230	0.530	0.650	0.740	----
14		0.080	0.370	0.089	0.030	0.290	0.390	0.470	0.750	----
119	Nickel copper steel plus molybdenum	0.030	0.160	0.009	0.030	0.010	0.530	0.290	----	0.08 Mo
65		0.130	0.450	0.073	0.020	0.066	0.573	0.730	----	0.087 Mo

Reference Copson Papers from ASTM Proceedings, Vols. 45, 48, 52, 60

Table II-J Average Weight Loss for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960)

Mat.	Weight Loss, g per sq. dm																
	Kure Beach (800 foot site), N. C. (marine)					Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID #	0.5 yr	1.5 yr	3.5 yr	7.5 yr	15.5 yr	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
25	4.6	gone	----	----	----	----	----	----	----	----	----	----	----	----	----	----	----
9	4.9	(b)	gone	----	----	----	----	----	----	----	----	----	----	----	----	----	----
44	6	(b)	gone	----	----	67(b)	gone	----	----	----	----	16.6	57.9(a)	gone at 4 yr	----	----	----
45	3.5	8.2	(b)	gone at 5 yr	----	18.8	39.7(a)	----	----	----	----	8.1	14.1	19.7	(c)	31.8(a)	----
63	2.3	4.8	10.1	19	43	6	17.4	27.1	gone	----	----	5.8	10.1	11.3	----	16.1	21.7
54	2.5	5	9.7	16.4	29.8	5.5	14.7	22.5(a)	gone	----	----	5.1	8.3	10.3	12.8	13.5	----
55	2.4	4.8	8.9	15	25.6	4.3	10.8	20.8(a)	27.8(b)	gone	----	4.4	7.5	8.6	10.4	10.8	----
10	2.6	5	10.3(a)	19.7(b)	(b)	6.5(a)	18.2(a)	gone	----	----	----	5.6	9.3	11.3	14.5	17.7	----
31	2.2	4.4	8.8	15.2	(b)	4.4	10.8(a)	17.2(a)	gone	----	----	5.3	8.6	10.1(a)	12.7(a)	15(a)	----
5	2.6	4.6	9.6(a)	16.5(b)	gone	5.3	11.3(a)	gone	----	----	----	5.2	8.6	10.9(a)	14(a)	15.8(a)	----
57	2.3	4.4	8.4	14.4	23.4	4	9.2	14.2	19.7(a)	(b)	----	4.5	7.1	8.8	9.9	10.6	----
6	2.5	4.6	8.8	14	22.9(a)	4.7	9.9	12.5(a)	17.7(b)	----	----	4.2	6.7	8.2	9.6	10.4(a)	----
41	2.1	3.9	8.2	13.9	23.4	3.7	----	13.7	19.3	23.4	----	4.1	----	8	----	10.3	14.8
43	2.2	4	8.1	13.7	23.1	3.8	----	(c)	17.2	22.2	----	4.1	----	8	----	10.5	14.7
64	2.2	4	7.9	13.6	22.1	3.9	----	13.5	----	21.5	----	3.9	----	7.6	----	10.2	14.3
40	2.1	4.2	8.2	13.8	23.6	3.9	----	16.6	21.4	25.8(a)	----	4.1	----	8.3	----	10.2	14.5
58	2.1	3.9	7.9	13.4	21.5	3.6	----	13.3	17.4	22.6	----	4	----	8.1	----	9.9	13.7
42	2.1	4.1	7.7	13.1	21.9	3.5	----	(c)	16.6	20.3	----	3.9	----	7.5	----	8.9	13.6
56	2.3	4.1	8	13.4	21.7	4.1	8.3	12.5	16.8(a)	21.5(a)	----	4.2	6.7	8	9.4	10	----
7	2.6	4.7	8.1	13.3	20.6(a)	4.1	8.7	12.1(a)	gone	----	----	4	6.2	6.9	8.3	9.4(a)	----
8	2.3	4.2	7	10.8	17.3(a)	----	----	----	----	----	----	----	----	----	----	----	----

Table II-J Average Weight Loss for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Mat.	Weight Loss, g per sq. dm																
	Kure Beach (800 foot site), N. C. (marine)					Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID #	0.5 yr	1.5 yr	3.5 yr	7.5 yr	15.5 yr	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
1	3.1	9.8(a)	gone	----	----	40.8(a)	gone	----	----	----	----	13.5	32(a)	gone	----	----	----
26	2.1	5.4	10.5(a)	17.7(b)	gone	----	----	----	----	----	----	----	----	----	----	----	----
11	2.1	4.2	8.1	13.5	20.9(a)	3.7	9.3(a)	15.7(a)	gone	----	----	4.9	7.9	9.7	12(a)	12.6(a)	----
2	2	3.8	8.1	10.8	17.9	3.4	----	10.3	----	15.9	28(a)	3.8	----	7	----	8.7	12
12	2	3.9	7.4	12	19(a)	3.4	7.3	9.8	13.4(a)	16.5(a)	----	3.4	5.7	6.6	8.3	9.1	----
16	2	3.8	6.9	10.9	16.9(a)	3.2	7.2	9.5	12.1(a)	14.3(a)	----	3.4	5.9	6.9	7.4	8.3	----
3	2	3.8	6.6	10.3	16.5	3.2	----	8.9	----	14.2	21.8	3.4	----	5.7	----	7.2	----
4	2.3	3.7	5.9	9	13.5	----	----	----	----	----	----	----	----	----	----	----	----
18	2.1	(c)	6.9	10.7	16.2(a)	3.3	7.2	10.9(a)	12.8(a)	16(a)	----	5.1	7.5	7.7	9.8	9.7(a)	----
19	1.8	3.7	6.4	10.6	16.9(a)	2.8	6.4	8.8	11.1(a)	14.3(a)	----	5.2	7.6	8.9	11.1	11.3	----
17	1.9	3.5	6.3	9.9	16.2(a)	3	6.3	8.3	10.9(a)	13.1(a)	----	3.9	5.7	6.3	(c)	8.3(a)	----
66	1.9	3.4	6.6	10.6	16.6	3	----	9		13.9	24.3	4.1	----	7.5	----	10.1	12.5
53	1.8	3.1	5.5	8.6	13.6	2.7	----	7.8	9.3	12.4	----	4	----	6.1	----	7.9	9.9
67	----	----	----	----	----	2.5	----	5.5	----	7.5	77.9	4.1	----	6.2	----	7.9	----
68	----	----	----	----	----	2.5	----	5.8	----	7.6	12	4.1	----	5.9	----	6.8	----
154	----	----	----	----	----	2.3	----	4.8	----	6.4	9.9	3	----	4.1	----	4.3	----
97	----	----	2.7	4.6	----	2.5	----	5.5	7.1	7.8	----	3.3	----	5.1	----	5.9	6.9
13	1.6	2.6	3.4	4.6	6.3	2.4	----	4.8	5.7	6.5	----	1.8	----	2.6	3.2	3.5	4.1
118	2.1	4.3	7.9	13.5	21.2	4.1	----	13.2	----	21.8	39.3(a)	4	----	7.7	----	9.4	13.4
120	2.3	4	7.7	12.7	20.2	3.6	----	12.9	----	21.1	35.9(a)	3.9	----	7.2	----	8.8	12.6
38	1.6	3	5.5	8.8	13.9	2.7	----	7.3	----	12.1	17.6	3.9	----	6.6	----	8.2	10.4
121	1.5	3.5	6.1	9.9	14.6	2.9	----	8.7	----	14	21.6	3	----	5.9	----	7.1	9.6
48	1.8	3.5	4.9	7.9	11.8	2.5	----	(c)	----	10.9	16.8	3	----	5.1	----	6.5	8.5

Table II-J Average Weight Loss for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Mat.	Weight Loss, g per sq. dm																
	Kure Beach (800 foot site), N. C. (marine)					Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID #	0.5 yr	1.5 yr	3.5 yr	7.5 yr	15.5 yr	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
32	2.1	3.9	7.4	13.1	21.5(a)	3.4	7.9	12.7(a)	17.2(a)	gone	----	4.7	7.2	8.8	10.5(a)	12.2(a)	----
27	1.7	3.4	6.2	10.3	15.9(a)	2.9	6.6	9.5(a)	11.3(a)	14.9(a)	----	4.3	6.1	7.2	8.8	9.7	----
33	1.9	4.1	7.7	12.6	20.4(a)	3.6	8.5	13.9(a)	gone	----	----	4.5	7.1	8	9.9	10.9(a)	----
34	1.7	3.1	5.5	8.6	14.2(a)	2.6	5.2	8.3	9.9(a)	12.8(a)	----	3.8	5.7	6.6	7.7	8.1	----
28	1.6	3	5.2	8.2	12.5(a)	2.4	5.2	6.8	8.3(a)	10.6(a)	----	3.3	4.6	5.8	6.9	7.9	----
147	----	----	----	----	----	2.3	4.8	6.2	7.8	9.5(a)	----	2.7	4.5	5.2	6.6	7.2	----
46	1.6	2.5	4.2	6.4	9.4	2.5	4.1	5.4	6.3	8.3	----	2.8	3.7	5.3	5.9	7	----
35	1.7	2.9	5.1	8	12.2	2.5	5	6.6	8.5	10.2	----	3.3	5.1	6.2	7	(c)	----
20	1.5	2.6	4.2	6.3	9.2	2	4	5.1	6.5	8.1	----	2.6	3.6	4.9	5.2	6.3	----
62	1.2	2	3	4.2	6.1	1.7	----	3.9	----	(c)	7.9	2.3	----	3.2	----	4.9	5.6
36	1.3	2.2	3.5	5.1	7.5	1.9	----	4.7	----	7.1	9.5	2.5	----	4.3	----	5.6	6.8
30	1.6	2.6	4.5	6.8	10.5(a)	2.3	4.4	5.5	6.8	(a)	----	3.7	4.4	5.1	5.9	6.4	----
29	1.4	2.8	4.3	6.5	9.8(a)	2.1	(c)	5.7	6.4	7.6	----	3	3.7	(c)	5.7	6.2	----
51	1.5	2.1	3.2	4.5	6.5	2	----	4.4	4.7	5.4	----	2.4	----	3.2	3.9	3.8	4.3
52	1.4	2.2	3.5	5	7.6	1.9	----	4.1	4.8	5.6	----	2.6	----	3.9	3.8	4.6	5.2
37	1.9	2.7	4.4	6.4	9.4	2.4	----	6.3	----	9.1	14.6	(c)	----	4.4	----	5.5	6.9
60	1.7	3	4.7	7.2	10.6	2.6	5.2	7.3	----	10.1	----	3.3	4.5	5.3	6.1	6.7	----
148	----	----	----	----	----	2.4	4.8	5.9	7	----	----	2.6	3.3	3.9	4.8	5.4	----
50	1.6	2.9	3.3	4.3	5.6	2.2	3.8	----	5.4	6.3	----	2.3	2.9	3.3	3.9	4.8	----
39	1.6	2.7	4.5	6.8	10	2.5	----	6.1	----	9.5	13.8	3.4	----	5.2	----	6.4	7.9
149	----	----	----	----	----	2.3	----	5.9	6.6	7.6	----	2.5	----	3.5	4.9	5.3	8.1
150	----	----	----	----	----	2.2	----	5.4	6.5	8	----	2.4	----	3.7	4.5	5	6.4
153	----	----	----	----	----	2	----	4.5	----	6.6	----	2.1	----	2.9	3.2	(c)	4.2
152	----	----	----	----	----	2	----	4.5	5.2	6.4	----	2.2	----	3.1	3.7	(c)	5.3
151	----	----	----	----	----	2	----	4.6	5.6	6.4	----	2	----	3	3.4	(c)	4.8

Table II-J Average Weight Loss for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

	Weight Loss, g per sq. dm																
Mat.	Kure Beach (800 foot site), N. C. (marine)					Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID #	0.5 yr	1.5 yr	3.5 yr	7.5 yr	15.5 yr	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
49	1.8	3.2	6.1	9.7	15.2	3.1	----	8.4		13.1	21.3	3.3	----	5.9		7.6	10.4
16-1	2	3.8	6.9	10.9	15.9(a)	3.4	6.6	10(a)	10.6(a)	14.5(a)	----	3.7	5.1	5.9	7	7.3	----
59	1.8	3.3	4.6	7	10.5	2.4	----	6.2	----	9.1	13.3	3.4	----	4.9	----	5.9	8.1
47	1.7	2.6	4.2	6.4	9.3	2.4	4.2	5.5	6.4	8.5	----	3.4	4	(c)	5	5.8	----
14	1.7	2.7	4.4	6.3	9.1	2.5	----	5.9		8.6	(c)	(c)	----	4.4	----	5.4	6.5
119	2	3.8	7	11.5	18.2	3.4	----	10.8	----	17	29.4	3.6	----	6.7	----	8	10.9
65	1.8	3.4	5.3	7.6	11.2	2.9	----	7.3	----	10.6	17.2	2.8	----	5.2	----	6.2	(c)

(a) Small perforations
(b) Large holes
(c) Attacked in cleaning
(d) Average corrosion rate 5.0 to 17.1 yr
(e) Average corrosion rate 7.1 to 18.1 yr

Reference Copson Papers from ASTM Proceedings, Vols. 45, 48, 52, 60

Table II-K Average Pit Depth for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960)

	Average pit depth, µm (average of 4 deepest pits on skyward surface and 4 deepest on groundward surface)											
Material	Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID Number	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
25	----	----	----	----	----	----	----	----	----	----	----	----
9	----	----	----	----	----	----	----	----	----	----	----	----
44	----	----	----	----	----	----	444.5	----	----	----	----	----
45	368.3	----	----	----	----	----	317.5	520.7	772.16	----	----	----
63	297.18	495.3	810.26	----	----	----	228.6	355.6	472.44	----	495.3	622.3
54	210.82	469.9	----	----	----	----	241.3	330.2	457.2	482.6	469.9	----
55	170.18	330.2	----	----	----	----	190.5	292.1	393.7	444.5	381	----
10	411.48(a)	----	----	----	----	----	228.6	342.9	500.38	523.24	558.8	----
31	182.88	----	----	----	----	----	215.9	292.1	393.7(a)	----	----	----
5	147.32	----	----	----	----	----	228.6	330.2	----	----	----	----
57	157.48	342.9	393.7	622.3(a)	----	----	203.2	279.4	386.08	381	419.1	----
6	157.48	279.4	----	----	----	----	177.8	241.3	327.66	388.62(a)	----	----
41	177.8	----	505.46	622.3	777.24	----	203.2	----	345.44	----	378.46	317.5
43	182.88	----	(c)	556.26	701.04	----	190.5	----	358.14	----	370.84	419.1
64	177.8	----	480.06	----	795.02	----	190.5	----	322.58	----	355.6	424.18
40	182.88	----	518.16	607.06	800.1(a)	----	203.2	----	320.04	----	383.54	421.64
58	177.8	----	528.32	548.64	688.34	----	177.8	----	373.38	----	370.84	406.4
42	190.5	----	469.9	622.3	635	----	165.1	----	322.58	----	375.92	370.84
56	190.5	304.8	469.9	----	----	----	190.5	292.1	347.98	373.38	378.46	----
7	134.62	279.4	----	----	----	----	177.8	254	325.12	350.52	----	----
8	----	----	----	----	----	----	----	----	----	----	----	----

Table II-K Average Pit Depth for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Material	Average pit depth, µm (average of 4 deepest pits on skyward surface and 4 deepest on groundward surface)											
	Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID Number	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
1	----	----	----	----	----	----	419.1	----	----	----	----	----
26	----	----	----	----	----	----	----	----	----	----	----	----
11	165.1	----	----	----	----	----	228.6	330.2	383.54	439.42(a)	----	----
2	203.2	----	383.54	----	574.04	----	165.1	----	304.8	----	355.6	330.2
12	160.02	254	411.48	----	----	----	165.1	215.9	337.82	360.68	332.74	----
16	147.32	254	312.42	419.1(a)	----	----	139.7	254	312.42	294.64	281.94	----
3	193.04	----	312.42	----	523.24	708.66	190.5	----	292.1	----	307.34	360.68
4	----	----	----	----	----	----	----	----	----	----	----	----
18	177.8	304.8	----	----	----	----	241.3	292.1	340.36	353.06	406.4(a)	----
19	165.1	279.4	439.42	----	----	----	254	279.4	373.38	363.22	363.22	----
17	152.4	254	335.28	----	----	----	203.2	228.6	337.82	340.36	368.3(a)	----
66	157.48	----	347.98	----	510.54	784.86	203.2	----	312.42	----	358.14	429.26
53	152.4	----	424.18	416.56	490.22	----	177.8	----	284.48	----	302.26	365.76
67	109.22	----	289.56	----	312.42	388.62	190.5	----	342.9	----	345.44	----
68	144.78	----	281.94	----	365.76	436.88	241.3	----	322.58	----	289.56	----
154	172.72	----	246.38	----	299.72	411.48	177.8	----	279.4	----	238.76	----
97	109.22	----	302.26	335.28	381	----	228.6	----	271.78	----	325.12	297.18
13	119.38	----	292.1	294.64	363.22	----	152.4	----	266.7	251.46	264.16	330.2
118	182.88	----	414.02	----	820.42	----	177.8	----	345.44	----	345.44	398.78
120	198.12	----	457.2	----	889	----	203.2	----	302.26	----	370.84	441.96
38	152.4	----	342.9	----	502.92	612.14	228.6	----	297.18	----	320.04	309.88
121	157.48	----	375.92	----	530.86	632.46	165.1	----	243.84	----	294.64	406.4
48	144.78	----	(c)	----	403.86	510.54	215.9	----	274.32	----	287.02	314.96

Table II-K Average Pit Depth for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

	Average pit depth, μm (average of 4 deepest pits on skyward surface and 4 deepest on groundward surface)											
Material	Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID Number	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
32	152.4	330.2	----	----	----	----	203.2	279.4	335.28	363.22(a)	393.7(a)	----
27	144.78	266.7	----	----	----	----	203.2	241.3	287.02	297.18	358.14	----
33	157.48	330.2	----	----	----	----	228.6	279.4	373.38	373.38	388.62(a)	----
34	167.64	254	403.86	431.8(a)	----	----	203.2	241.3	309.88	332.74	330.2	----
28	119.38	228.6	337.82	355.6(a)	----	----	228.6	215.9	284.48	304.8	264.16	----
147	147.32	215.9	284.48	342.9	381(a)	----	190.5	203.2	274.32	287.02	266.7	----
46	127	215.9	256.54	350.52	332.74	----	177.8	190.5	264.16	251.46	256.54	----
35	134.62	152.4	246.38	312.42	345.44	----	241.3	203.2	248.92	259.08	(c)	----
20	119.38	139.7	215.9	299.72	337.82	----	203.2	165.1	233.68	248.92	228.6	----
62	93.98	----	165.1	----	(c)	299.72	127	----	226.06	----	185.42	246.38
36	127	----	132.08	----	274.32	327.66	139.7	----	198.12	----	205.74	289.56
30	109.22	215.9	289.56	375.92	----	----	254	228.6	322.58	302.26	289.56	----
29	127	(c)	226.06	360.68	439.42	----	177.8	203.2	(c)	226.06	269.24	----
51	127	----	236.22	256.54	340.36	----	190.5	----	274.32	228.6	269.24	274.32
52	106.68	----	213.36	276.86	355.6	----	177.8	----	236.22	261.62	243.84	292.1
37	121.92	----	304.8	----	381	467.36	152.4	----	246.38	----	261.62	274.32
60	134.62	228.6	274.32	----	406.4	----	203.2	228.6	292.1	279.4	274.32	----
148	114.3	203.2	246.38	373.38	----	----	177.8	165.1	274.32	248.92	279.4	----
50	109.22	165.1	----	248.92	292.1	----	127	203.2	210.82	236.22	264.16	----
39	177.8	----	269.24	----	368.3	561.34	203.2	----	248.92	----	274.32	307.34
149	177.8	----	284.48	332.74	396.24	----	114.3	----	205.74	208.28	251.46	342.9
150	127	----	251.46	337.82	355.6	----	177.8	----	251.46	233.68	243.84	314.96
153	93.98	----	261.62	271.78	294.64	----	215.9	----	210.82	218.44	254	327.66
152	127	----	274.32	264.16	304.8	----	177.8	----	226.06	238.76	218.44	269.24
151	132.08	----	210.82	261.62	327.66	----	139.7	----	203.2	226.06	276.86	289.56

Table II-K Average Pit Depth for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Material	Average pit depth, µm (average of 4 deepest pits on skyward surface and 4 deepest on groundward surface)											
	Block Island, R. I. (marine)						Bayonne, N. J. (industrial)					
ID Number	1.1 yr	3.3 yr	5.0 yr	7.2 yr	9.1 yr	17.1 yr	1.0 yr	3.0 yr	5.1 yr	7.1 yr	9.1 yr	18.1 yr
49	182.88	----	314.96	----	452.12	685.8	203.2	----	281.94	----	294.64	320.04
16-1	165.1	279.4	406.4(a)	----	----	----	228.6	----	299.72	302.26	312.42	----
59	152.4		274.32	----	459.74	505.46	241.3	----	307.34	----	287.02	363.22
47	170.18	241.3	284.48	327.66	393.7	----	203.2	241.3	(c)	264.16	294.64	----
14	152.4	----	289.56	----	431.8	----	(c)	----	264.16	----	312.42	314.96
119	147.32	----	373.38	----	713.74	899.16	152.4	----	309.88	----	342.9	381
65	152.4	----	279.4	----	431.8	497.84	165.1	----	236.22	----	274.32	(c)

(a) Small perforations
(b) Large holes
(c) Attacked in cleaning
(d) Average corrosion rate 5.0 to 17.1 yr
(e) Average corrosion rate 7.1 to 18.1 yr

Reference Copson Papers from ASTM Proceedings, Vols. 45, 48, 52, 60

Table II-L Average Corrosion Rate for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960)

Material ID Number	Average Corrosion Rate Calculated From Specimen Weight Loss, µm/yr								
	Kure Beach, N. C. (marine) (800 Foot site)			Block Island, R. I. (marine)			Bayonne, N. J. (industrial)		
	0 to 7.5 yr	0 to 17.5 yr	7.5 to 15.5 yr	0 to 9.1 yr	0 to 17.1 yr	9.1 to 17.1 yr	0 to 9.1 yr	0 to 18.1 yr	9.1 to 18.5 yr
25	----	----	----	----	----	----	----	----	----
9	----	----	----	----	----	----	----	----	----
44	----	----	----	----	----	----	----	----	----
45	----	----	----	----	----	----	43.434	----	----
63	32.766	35.56	38.354	----	----	----	22.606	15.494	7.874
54	28.194	24.638	21.59	----	----	----	19.05	----	----
55	25.654	21.336	17.018	----	----	----	15.24	----	----
10	33.782	----	----	----	----	----	24.892	----	----
31	25.908	----	----	----	----	----	21.082	----	----
5	28.194	----	----	----	----	----	22.098	----	----
57	24.638	19.304	14.478	----	----	----	14.986	----	----
6	23.876	19.05	14.224	----	----	----	14.478	----	----
41	23.622	19.304	15.24	33.02	----	----	14.478	10.414	6.35
43	23.368	19.05	14.986	31.242	----	----	14.732	10.414	6.096
64	23.114	18.288	13.716	30.48	----	----	14.224	10.16	5.842
40	23.622	19.558	15.748	36.322	----	----	14.224	10.16	6.096
58	22.86	17.78	12.954	32.004	----	----	13.97	9.652	5.334
42	22.352	18.034	14.224	28.448	----	----	12.446	9.652	6.858
56	22.86	18.034	13.208	30.226	----	----	13.97	----	----
7	22.606	17.018	11.684	----	----	----	13.208	----	----
8	18.542	14.224	10.414	----	----	----	----	----	----

Table II-L Average Corrosion Rate for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Material ID Number	Average Corrosion Rate Calculated From Specimen Weight Loss, µm/yr								
	Kure Beach, N. C. (marine) (800 Foot site)			Block Island, R. I. (marine)			Bayonne, N. J. (industrial)		
	0 to 7.5 yr	0 to 17.5 yr	7.5 to 15.5 yr	0 to 9.1 yr	0 to 17.1 yr	9.1 to 17.1 yr	0 to 9.1 yr	0 to 18.1 yr	9.1 to 18.5 yr
1	----	----	----	----	----	----	----	----	----
26	30.226	----	----	----	----	----	----	----	----
11	23.114	17.272	11.938	----	----	----	17.78	----	----
2	18.542	14.732	11.43	22.352	21.082	19.558	12.192	8.382	4.826
12	20.574	15.748	11.176	23.114	----	----	12.7	----	----
16	18.542	13.97	9.652	20.066	----	----	11.684	----	----
3	17.526	13.716	7.366	19.812	16.256	12.192	10.16	7.112	4.318
4	15.494	11.176	7.112	----	----	----	----	----	----
18	18.288	13.462	8.89	22.352	----	----	13.716	----	----
19	18.034	13.97	10.16	20.066	----	----	15.748	----	----
17	17.018	13.462	10.16	18.288	----	----	11.684	----	----
66	18.034	13.716	9.652	19.558	18.288	16.764	14.224	8.89	3.556
53	14.732	11.176	8.128	17.272	----	----	11.176	7.112	2.794
67	----	----	----	10.668	8.89	7.112	11.176	----	----
68	----	----	----	10.668	8.89	7.112	9.652	----	----
154	----	----	----	8.89	7.366	5.588	6.096	----	----
97	7.874	----	----	10.922	----	----	8.382	4.826	1.524
13	7.874	5.334	2.794	9.144	----	----	4.826	2.794	0.762
118	23.114	17.526	12.446	30.734	29.464	27.94	13.208	9.398	5.588
120	21.844	16.764	11.938	29.718	26.924	23.876	12.446	8.89	5.588
38	14.986	11.43	8.128	17.018	13.208	8.89	11.43	7.366	3.048
121	17.018	12.192	7.62	19.558	16.256	12.192	9.906	6.858	3.556
48	13.462	9.652	6.35	15.24	12.7	9.398	9.144	6.096	2.794

Table II-L Average Corrosion Rate for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Material ID Number	Average Corrosion Rate Calculated From Specimen Weight Loss, µm/yr								
	Kure Beach, N. C. (marine) (800 Foot site)			Block Island, R. I. (marine)			Bayonne, N. J. (industrial)		
	0 to 7.5 yr	0 to 17.5 yr	7.5 to 15.5 yr	0 to 9.1 yr	0 to 17.1 yr	9.1 to 17.1 yr	0 to 9.1 yr	0 to 18.1 yr	9.1 to 18.5 yr
32	22.352	17.78	13.462	----	----	----	17.018	----	----
27	17.526	13.208	8.89	20.828	----	----	13.716	----	----
33	21.59	16.764	12.446	----	----	----	15.24	----	----
34	14.732	11.684	8.89	17.78	----	----	11.43	----	----
28	13.97	10.414	6.858	14.986	----	----	11.176	----	----
147	----	----	----	13.462	----	----	10.16	----	----
46	10.922	10.414	9.906	11.684	----	----	9.906	----	----
35	13.716	10.16	6.858	14.224	----	----	----	----	----
20	10.668	7.62	4.572	11.43	----	----	8.89	----	----
62	7.112	5.08	3.048	----	5.842	4.318(d)	6.858	4.064	1.016
36	8.636	6.096	3.81	9.906	7.112	3.81	7.874	4.826	1.778
30	11.684	8.636	5.842	----	----	----	225.806	----	----
29	11.176	8.128	5.334	10.668	----	----	219.3544	----	----
51	7.62	5.334	3.302	7.62	----	----	5.588	3.048	0.762
52	8.636	6.35	4.064	7.874	----	----	6.35	3.556	0.762
37	10.922	7.874	4.826	12.7	10.922	8.89	7.62	4.826	2.032
60	12.446	8.89	5.334	14.224	----	----	9.398	----	----
148	----	----	----	----	----	----	7.62	----	----
50	7.366	4.572	2.032	8.636	----	----	6.858	----	----
39	11.684	8.382	5.08	13.462	10.414	6.858	8.89	5.588	2.032
149	----	----	----	10.668	----	----	7.366	5.842	4.064
150	----	----	----	11.176	----	----	7.112	4.572	2.032
153	----	----	----	9.398	----	----	----	3.048	1.27(e)
152	----	----	----	8.89	----	----	----	3.81	1.778(e)
151	----	----	----	8.89	----	----	----	3.302	1.524(e)

Table II-L Average Corrosion Rate for Iron and Steel Specimens Exposed to Industrial and Marine Environments for Periods of 15.5 to 18.1 Years

(Copson; 1945, 1948, 1952, 1960) -- (Continued)

Material	Average Corrosion Rate Calculated From Specimen Weight Loss, µm/yr								
	Kure Beach, N. C. (marine) (800 Foot site)			Block Island, R. I. (marine)			Bayonne, N. J. (industrial)		
ID Number	0 to 7.5 yr	0 to 17.5 yr	7.5 to 15.5 yr	0 to 9.1 yr	0 to 17.1 yr	9.1 to 17.1 yr	0 to 9.1 yr	0 to 18.1 yr	9.1 to 18.5 yr
49	16.51	12.7	8.89	18.288	16.002	13.208	10.668	7.366	4.064
16-1	18.542	13.208	8.128	20.32	----	----	10.16	----	----
59	11.938	8.636	5.588	12.7	9.906	6.858	8.382	5.842	3.048
47	10.922	7.62	4.572	11.938	----	----	8.128	----	----
14	10.668	7.62	4.572	12.192	----	----	7.62	4.572	1.524
119	19.558	14.986	10.668	23.876	22.606	18.542	11.176	7.62	4.064
65	12.954	9.398	5.842	14.986	12.954	10.668	8.636	----	----

(a) Small perforations
(b) Large holes
(c) Attacked in cleaning
(d) Average corrosion rate 5.0 to 17.1 yr
(e) Average corrosion rate 7.1 to 18.1 yr

Reference Copson Papers from ASTM Proceedings, Vols. 45, 48, 52, 60

Table II-M Composition of Metals Exposed in Tropical and Temperate Climates (Southwell, et al., 1958; 1960; 1976)

Metal	Type	Specification	C	Mn	P	S	Si	Cr	Ni	Cu	Mo
A	Unalloyed low-carbon	QQ-S-741, Type II Grade A, Class 1	0.250	0.460	0.083	0.014	0.004	0.025	0.035	0.020	...
B*	Unalloyed low-carbon		0.160	0.420	0.013	0.021	0.010	0.010	0.020	0.020	...
C	Low-alloy	Proprietary Cu-Ni	0.130	0.410	0.007	0.026	0.048	...	1.700	0.900	...
D	Low-alloy	Proprietary Cu-Cr-Si	0.120	0.410	0.084	0.026	0.500	0.600	0.630	0.430	...
E	Low-alloy	Proprietary Cu-Ni-Mn-Mo	0.087	0.730	0.650	0.018	0.067	0.030	0.820	0.590	0.190
F	Low-alloy	Proprietary Cr-Ni-Mn	0.110	0.540	0.086	0.029	0.180	0.500	0.400	0.570	...
G*	Low-alloy		0.090	0.240	0.150	0.024	0.800	1.100	0.050	0.430	...
H*	Low-alloy		0.060	0.480	0.110	0.030	0.540	1.000	0.510	0.410	...
I*	Low-alloy		0.050	0.360	0.050	0.016	0.008	0.010	2.000	1.100	...
J*	Low-alloy		0.110	0.550	0.080	0.026	0.060	0.310	0.280	0.550	...
K*	Low-alloy		0.160	1.400	0.013	0.021	0.180	0.030	0.500	0.300	...
L*	Copper-bearing		0.270	0.480	0.024	0.036	0.009	0.120	0.010	0.210	...
M	Copper-bearing	QQ-S-741 Class 2	0.220	0.410	0.004	0.030	0.008	0.015	0.100	0.240	...
N	Nickel (2%)	RR-SPECS-3-A	0.180	0.520	0.036	0.024	0.190	0.140	2.060	0.610	...
O	Nickel (5%)	SAE-2515 5% Ni Steel	0.100	0.500	0.010	0.210	0.200	0.070	4.550	0.054	0.074
P	Chromium (3%)	Max. 0.10%C Hot Rolled	0.070	0.540	0.008	0.016	0.140	3.240	0.130	0.110	0.032
Q	Chromium (5%)	Co 1% AISI Type 501D	0.120	0.580	0.020	0.014	0.320	5.200	0.060	0.036	0.410
R	Wrought Iron	ASTM A162-39	0.020	0.024	0.117	0.015	0.095	0.010	0.007	0.018	...

The metals with an "*" were exposed to a temperate climate; others were exposed to a tropical climate.

Table III-A Composition of Structural Ferrous Materials Tested in Tropical Waters

(Southwell and Alexander, 1970)

KEY	Metal	Surface Condition	C	Mn	P	S	Si	Cr	Ni	Cu	Mo
A	C Steel	Pickled	0.24	0.48	0.04	0.027	0.008	0.03	0.051	0.08	--
B	C Steel	Machined	0.24	0.48	0.04	0.027	0.008	0.03	0.051	0.08	--
C	C Steel	Millscale	0.24	0.48	0.04	0.027	0.008	0.03	0.051	0.08	--
D	C Steel - 0.3% Ni	Pickled	0.22	0.44	0.019	0.033	0.009	tr.	0.14	0.35	--
E	2% Ni Steel	Pickled	0.2	0.54	0.012	0.023	0.18	0.15	1.94	0.63	--
F	5% Ni Steel	Pickled	0.13	0.49	0.01	0.014	0.16	0.1	5.51	0.062	--
G	3% Cr Steel	Pickled	0.08	0.44	0.01	0.017	0.13	3.16	0.16	0.11	0.02
H	5% Cr steel	Pickled	0.08	0.41	0.02	0.019	0.2	5.06	0.11	0.062	0.52
I	Low Alloy Steel	Pickled	0.08	0.47	0.007	0.026	0.06	--	1.54	0.87	--
J	Low Alloy Steel	Pickled	0.15	0.45	0.113	0.026	0.47	0.68	0.49	0.42	--
K	Low Alloy Steel	Pickled	0.078	0.75	0.058	0.022	0.04	tr.	0.72	0.61	0.13
L	Low Alloy Steel	Pickled	0.13	0.6	0.089	0.021	0.15	0.55	0.3	0.61	0.059
M	Cast Steel	Machined	0.27	0.68	0.028	0.028	0.41	0.12	0.22	0.1	--
N	Cast Steel	As Received	0.27	0.68	0.028	0.028	0.41	0.12	0.22	0.1	--
O	Gray Cast Iron	Machined	3.18	0.8	0.162	0.103	1.98	0.57	0.31	0.08	--
P	Gray Cast Iron	As Received	3.18	0.8	0.162	0.103	1.98	0.57	0.31	0.08	--
Q	Austenitic Cast Iron	Machined	2.66	0.94	0.24	0.104	3.17	2.29	17.9	0.8	--
R	Austenitic Cast Iron	As Recieved	2.66	0.94	0.24	0.104	3.17	2.29	17.9	0.8	--
S	Wrought Iron	Machined	0.04	0.038	0.141	0.018	0.098	tr.	0.006	0.02	--
T	Wrought Iron	As Received	0.04	0.038	0.141	0.018	0.098	tr.	0.006	0.02	--

C.R. Southwell and A.L. Alexander, "Corrosion of Metals In Tropical Waters Structural Ferrous Materials," in Materials Protection, January 1970, p. 14

Table III-B Corrosion to Structural Ferrous Metal Continuously Immersed In Fresh Water
Gatun Lake, Panama Canal Zone
(Southwell & Alexander, 1970)

			Weight Loss (mdd)					Corrosion Rate (µm/yr)					Average 20 Deepest Pits (µm)			Deepest Pit (µm)		
KEY	Metal	Surface Condition	1 yr	2 yr	4 yr	8 yr	16 yr	1 yr	2 yr	4 yr	8 yr	16 yr	1 yr	8 yr	16 yr	1 yr	8 yr	16 yr
A	C Steel	Pickled	41.6	32.5	23.3	15.0	9.6	195.6	152.4	108.0	69.9	44.5	508.0	1473.2	1828.8	609.6	1803.4	2362.2
B	C Steel	Machined	41.1	31.8	21.8	13.8	8.4	190.5	152.4	101.6	63.5	39.7	558.8	1524.0	1651.0	660.4	2235.2	2387.6
C	C Steel	Millscale	34.5	26.0	18.6	12.9	8.7	160.0	120.7	88.9	60.3	39.7	736.6	1219.2	1676.4	1016.0	1879.6	2311.4
D	C Steel - 0.3% Ni	Pickled	43.0	33.2	24.1	15.9	9.8	200.7	152.4	114.3	73.0	46.0	558.8	1625.6	1625.6	863.6	2082.8	2260.6
E	2% Ni Steel	Pickled	40.3	34.7	23.8	13.8	7.5	188.0	165.1	108.0	63.5	34.9	482.6	1600.2	1676.4	736.6	2260.6	2209.8
F	5% Ni Steel	Pickled	35.9	33.7	24.3	16.0	10.2	167.6	152.4	114.3	76.2	46.0	457.2	1752.6	1803.4	838.2	2540.0	2514.6
G	3% Cr Steel	Pickled	13.3	10.0	7.4	6.6	5.0	61.0	47.0	34.3	30.8	23.8	609.6	1371.6	1955.8	711.2	1600.2	2489.2
H	5% Cr steel	Pickled	10.4	6.2	4.9	5.9	4.6	48.3	29.2	22.9	27.3	20.6	584.2	1320.8	1701.8	736.6	1473.2	2438.4
I	Low Alloy Steel	Pickled	33.2	27.0	17.8	11.2	7.5	152.4	125.7	82.6	50.8	34.9	431.8	1701.8	1905.0	787.4	2159.0	2794.0
J	Low Alloy Steel	Pickled	31.0	23.6	16.2	11.3	8.3	144.8	109.2	76.2	54.0	38.1	1016.0	2006.6	2235.2	1422.4	2362.2	2844.8
K	Low Alloy Steel	Pickled	39.7	31.0	21.7	15.3	10.6	185.4	139.7	101.6	69.9	49.2	609.6	2057.4	2565.4	812.8	3048.0	3149.6
L	Low Alloy Steel	Pickled	29.3	22.3	17.3	10.7	9.1	137.2	104.1	82.6	50.8	42.9	1016.0	2819.4	2870.2	1193.8	3708.4	3581.4
S	Wrought Iron	Machined	41.1	31.6	19.4	9.9	ns	193.0	152.4	88.9	47.6	ns	685.8	1270.0	ns	812.8	2260.6	ns
T	Wrought Iron	As Received	40.0	30.7	20.2	12.7	ns	188.0	139.7	95.3	60.3	ns	457.2	1371.6	ns	584.2	2032.0	ns
M	Cast Steel	Machined	44.4	32.5	19.7	13.1	9.0	208.3	152.4	88.9	60.3	41.3	762.0	1701.8	2489.2	965.2	2336.8	3556.0
N	Cast Steel	As Received	37.5	29.7	19.9	nd	nd	193.0	152.4	101.6	nd	nd	1574.8	nd	nd	2946.4	nd	nd
O	Gray Cast Iron	Machined	34.2	28.8	18.5	14.2	10.1	177.8	152.4	95.3	73.0	52.4	1320.8	2692.4	2743.2	1727.2	3327.4	perf
P	Gray Cast Iron	As Received	38.9	31.9	20.4	nd	nd	200.7	165.1	108.0	nd	nd	1828.8	nd	nd	3810.0	nd	nd
Q	Austenitic Cast Iron	Machined	4.5	3.9	3.5	2.6	2.5	22.9	20.3	17.8	13.3	12.7	228.6	685.8	939.8	279.4	914.4	1549.4

Reference :
Southwell and Alexander "Corrosion of Metals In Tropical Waters Structural Ferrous Metals," in Materials Protection January 1970.

Table III-C Corrosion to Structural Ferrous Metal Continuously Immersed In Sea Water Pacific Ocean, Panama Canal Zone

(Southwell & Alexander, 1970)

			Weight Loss (mdd)					Corrosion Rate (µm/yr)					Average 20 Deepest Pits (µm)			Deepest Pit (µm)		
KEY	Metal	Surface Condition	1 yr	2 yr	4 yr	8 yr	16 yr	1 yr	2 yr	4 yr	8 yr	16 yr	1 yr	8 yr	16 yr	1 yr	8 yr	16 yr
A	C Steel	Pickled	32.3	22.6	21.5	17.4	15.5	149.9	105.4	101.6	82.6	73.0	1041.4	1676.4	2286.0	1574.8	2184.4	3937.0
B	C Steel	Machined	30.1	24.7	20.4	18.5	16.9	139.7	114.3	95.3	85.7	77.8	965.2	1473.2	perf	1625.6	3987.8	perf
C	C Steel	Millscale	23.0	21.2	20.6	15.9	perf	106.7	99.1	95.3	73.0	perf	1397.0	perf	perf	2641.6	perf	perf
D	C Steel - 0.3% Ni	Pickled	32.3	25.1	19.9	18.9	16.8	149.9	116.8	95.3	88.9	77.8	914.4	1600.2	2159.0	1549.4	2743.2	3149.6
E	2% Ni Steel	Pickled	40.5	25.1	23.4	21.6	17.8	188.0	116.8	108.0	101.6	82.6	76.2	2387.6	perf	1295.4	4546.6	perf
F	5% Ni Steel	Pickled	34.2	27.4	22.7	21.8	17.6	160.0	127.0	108.0	101.6	82.6	736.6	2971.8	perf	1447.8	5435.6	perf
G	3% Cr Steel	Pickled	11.3	14.9	24.4	27.5	24.0	53.3	69.9	114.3	127.0	111.1	279.4	1651.0	2362.2	508.0	1981.2	2768.6
H	5% Cr steel	Pickled	14.6	16.6	23.0	21.7	20.4	68.6	77.5	108.0	101.6	88.9	685.8	1600.2	1752.6	2311.4	2286.0	2540.0
I	Low Alloy Steel	Pickled	32.1	23.2	28.2	18.0	16.6	149.9	108.0	133.4	82.6	76.2	1371.6	2082.8	perf	1651.0	3860.8	perf
J	Low Alloy Steel	Pickled	37.3	30.5	25.7	29.3	27.4	175.3	139.7	120.7	136.5	127.0	787.4	2032.0	perf	1168.4	4445.0	perf
K	Low Alloy Steel	Pickled	35.1	26.2	20.7	17.4	15.4	162.6	121.9	95.3	79.4	71.4	685.8	1422.4	1803.4	1193.8	3530.6	2794.0
L	Low Alloy Steel	Pickled	34.5	30.0	28.2	29.9	28.6	160.0	139.7	133.4	139.7	133.4	660.4	perf	perf	914.4	perf	perf
S	Wrought Iron	Machined	34.8	20.7	19.3	15.0	ns	165.1	97.8	88.9	69.9	ns	863.6	1600.2	ns	1295.4	3505.2	ns
T	Wrought Iron	As Received	34.2	25.2	20.5	17.6	14.8	160.0	118.1	95.3	82.6	69.9	1168.4	perf	5080.0	1854.2	perf	5080.0
M	Cast Steel	Machined	40.3	25.6	20.5	18.9	14.7	188.0	119.4	95.3	88.9	68.3	1422.4	2133.6	3708.4	2133.6	3429.0	3708.4
N	Cast Steel	As Received	34.5	16.3	10.8	nd	nd	160.0	76.2	50.2	nd	nd	1651.0	nd	nd	3327.4	nd	nd
O	Gray Cast Iron	Machined	48.8	37.7	27.1	35.6	29.1	251.5	203.2	139.7	184.2	150.8	1066.8	2463.8	nd	1473.2	4064.0	nd
P	Gray Cast Iron	As Received	31.0	19.3	18.3	nd	nd	160.0	100.3	95.3	nd	nd	1752.6	nd	nd	5181.6	nd	nd
Q	Austenitic Cast Iron	Machined	16.4	12.3	9.9	9.3	7.1	83.8	63.5	50.8	47.6	36.5	127.0	1066.8	2946.4	127.0	2895.6	2946.4
R	Austenitic Cast Iron	As Received	12.4	5.5	nd	nd	nd	63.5	27.9	nd	nd	nd	127.0	nd	nd	127.0	nd	nd

Reference : Southwell and Alexander "Corrosion of Metals In Tropical Waters Structural Ferrous Metals," in Materials Protection January 1970

Table III-D Corrosion to Structural Ferrous Metal Continuously Mean Tide Elevation

Pacific Ocean, Panama Canal Zone

(Southwell & Alexander, 1970)

		Surface	Weight Loss (mdd)					Corrosion Rate (μm/yr)					Average 20 Deepest Pits (μm)			Deepest Pit (μm)		
KEY	Metal	Condition	1 yr	2 yr	4 yr	8 yr	16 yr	1 yr	2 yr	4 yr	8 yr	16 yr	1 yr	8 yr	16 yr	1 yr	8 yr	16 yr
A	C Steel	Pickled	51.5	25.3	17.4	15.9	16.2	241.3	118.1	82.6	73.0	74.6	406.4	1016.0	1676.4	685.8	1651.0	2489.2
B	C Steel	Machined	57.5	29.7	20.9	17.0	14.3	279.4	139.7	95.3	79.4	66.7	533.4	1143.0	1320.8	1168.4	3149.6	1854.2
C	C Steel	Millscale	50.4	30.8	18.0	12.6	14.9	236.2	139.7	82.6	57.2	69.9	457.2	1092.2	1371.6	609.6	1905.0	2590.8
D	C Steel - 0.3% Ni	Pickled	61.4	31.0	15.8	16.5	15.8	279.4	139.7	76.2	76.2	73.0	609.6	1143.0	1295.4	965.2	1600.2	1701.8
E	2% Ni Steel	Pickled	48.5	24.5	18.8	15.6	18.8	226.1	114.3	88.9	73.0	87.3	558.8	990.6	1778.0	990.6	1270.0	2108.2
F	5% Ni Steel	Pickled	42.7	23.6	17.4	13.6	18.3	198.1	110.5	82.6	63.5	85.7	nd	990.6	1574.8	nd	1905.0	2311.4
G	3% Cr Steel	Pickled	60.5	32.3	19.9	17.5	16.4	279.4	152.4	95.3	79.4	76.2	584.2	2082.8	perf	812.8	2362.2	perf
H	5% Cr steel	Pickled	63.8	31.0	17.1	16.6	16.1	304.8	152.4	82.6	79.4	74.6	711.2	2235.2	perf	939.8	2514.6	perf
I	Low Alloy Steel	Pickled	64.9	35.6	20.5	27.2	16.8	304.8	165.1	95.3	127.0	77.8	nd	1778.0	1473.2	nd	3403.6	2362.2
J	Low Alloy Steel	Pickled	58.6	26.7	16.6	14.3	16.8	279.4	127.0	76.2	66.7	77.8	508.0	1193.8	1524.0	685.8	1371.6	1828.8
K	Low Alloy Steel	Pickled	59.2	41.5	25.5	16.9	20.9	279.4	190.5	120.7	79.4	96.8	609.6	1016.0	1498.6	812.8	2387.6	3733.8
L	Low Alloy Steel	Pickled	49.9	24.7	14.7	14.0	15.6	231.1	114.3	69.9	66.7	73.0	482.6	990.6	1549.4	685.8	1270.0	2362.2
S	Wrought Iron	Machined	39.2	21.4	16.4	11.1	ns	185.4	100.3	76.2	50.8	ns	431.8	889.0	ns	685.8	1244.6	ns
T	Wrought Iron	As Received	32.6	16.8	11.0	11.6	12.2	154.9	78.7	52.1	54.0	57.2	431.8	889.0	nd	584.2	1041.4	nd
M	Cast Steel	Machined	43.3	20.8	14.9	14.5	16.7	200.7	97.8	69.9	66.7	77.8	889.0	1346.2	2311.4	1955.8	1930.4	4038.6
N	Cast Steel	As Received	29.0	18.6	13.2	nd	nd	134.6	86.4	61.0	nd	nd	1651.0	nd	nd	6502.4	nd	nd
O	Gray Cast Iron	Machined	126.0	72.1	37.8	28.0	26.9	660.4	368.3	196.9	146.1	138.1	1397.0	perf	perf	2032.0	perf	nd
P	Gray Cast Iron	As Received	33.2	22.7	11.5	nd	nd	170.2	116.8	59.1	nd	nd	1473.2	nd	nd	3175.0	nd	nd
Q	Austenitic Cast Iron	Machined	8.1	5.1	3.7	2.4	2.3	40.6	26.7	19.1	12.4	11.6	nd	1066.8	1701.8	nd	2895.6	2895.6
R	Austenitic Cast Iron	As Received	3.8	1.6	1.5	nd	nd	20.3	7.6	7.6	nd	nd	nd	nd	nd	nd	nd	nd

Reference : Southwell and Alexander "Corrosion of Metals In Tropical Waters Structural Ferrous Metals," in Materials Protection January 1970

Table III-E Corrosion Performance Comparison Between Carbon Steel and Austenitic Cast Iron in Three Aqueous Environments (Southwell & Alexander, 1970)

	Carbon Steel		Austenitic Cast Iron	
Environment	General µm/yr	Pitting µm	General µm/yr	Pitting µm
Fresh water	39.7	2388	12.7	1549
Sea water immersion	77.8	Perforated	36.5	2947
Mean Tide	66.7	1854	11.6	2896

Table III-F Corrosion of Structural Steel At Kure Beach, NC

(Larrabee, 1953)

Key	Metal	Composition (%)									Average Penetration Rate (mm/yr)		
		C	Mn	P	S	Si	Cr	Ni	Cu	Mo	1.5 yr	2.5 yr	4.5 yr
U	Cor-Ten	0.08	0.36	0.08	0.03	0.31	0.73	0.49	0.41	--	106.7	109.2	96.5
V	Tri-Ten	0.17	1.20	0.03	0.03	0.18	0.07	0.53	0.31	--	111.8	96.5	76.2
W	Ni-Cu	0.19	0.44	0.03	0.03	0.03	0.22	1.80	0.81	--	134.6	114.3	88.9
X	Cr-Mo	0.07	0.39	0.01	0.02	0.09	2.60	0.04	0.02	0.52	35.6	40.6	40.6
Y	Structural Carbon	0.25	0.42	0.02	0.03	0.03	0.06	0.02	0.01	--	121.9	104.1	83.8

Reference:

C.P. Larrabee, "Corrosion Resistance of High-Strength Low-Alloy Steels as Influenced by Composition and Environment," in Corrosion Vol 9, #8, p 259

Table III-G Corrosion of Low-Alloy Steels at Emsworth, England

(Hudson & Stanners, 1955)

				Composition (%)						Average Penetration Rate (µm/yr)	
Key	C	Si	Mn	S	P	Ni	Cu	Cr	Other	2 yr	5 yr
Basis 0.1% Carbon Steel											
1	0.08	0.18	0.65	0.047	0.027	tr	0.014	0.01		69.9	56.4
6	0.04	0.21	0.48	0.052	0.057		0.061			71.1	72.1
27	0.13	0.27	0.69	0.043	0.048		0.017			68.6	61.5
36	0.05	0.17	0.34	0.012	0.019		0.01	0.01		71.1	74.2
47	0.06	0.13	0.47	0.04	0.015	0.09	0.23	0.04		59.7	63.5
Effect of Carbon Content											
1	0.08	0.18	0.65	0.047	0.027	tr	0.014	0.01		69.9	56.4
2	0.12	0.17	0.62	0.04	0.028	tr	0.015	0.01		71.1	80.8
3	0.29	0.17	0.76	0.043	0.027	tr	0.014	0.03		68.6	67.6
4	0.51	0.18	0.78	0.031	0.026	tr	0.016	0.01		76.2	74.7
36	0.05	0.17	0.34	0.012	0.019		0.01	0.01		71.1	74.2
37	0.12	0.15	0.44	0.011	0.021		0.014	0.05		63.5	58.4
38	0.26	0.09	0.58	0.012	0.024		0.018			68.6	67.1
39	0.39	0.17	0.59	0.012	0.026		0.016	0.02		81.3	75.2
Effect of Manganese Content											
2	0.12	0.17	0.62	0.04	0.028	tr	0.015	0.01		71.1	80.8
5	0.18	0.11	1.82	0.047	0.033	tr	0.014	0.01		76.2	64.0
8	0.2	0.22	0.52	0.053	0.057		0.52			63.5	71.1
7	0.17	0.12	1.71	0.045	0.032	tr	0.52	0.01		66.0	64.0
Effect of Chromium, Copper, Phosphorus, and Silicon Additions											
6	0.04	0.21	0.48	0.052	0.057		0.061			71.1	72.1
11	0.06	0.81	0.54	0.054	0.057		0.06			67.3	53.3
12	0.09	0.17	0.52	0.051	0.164		0.059			71.1	60.5
18	0.11	0.8	0.49	0.053	0.053		0.53			61.0	72.6
13	0.1	0.2	0.58	0.05	0.156		0.48			67.3	59.9
9	0.12	0.2	0.54	0.054	0.049		0.059	1.04		52.1	64.5
17	0.1	0.83	0.59	0.049	0.058		0.057	1.04		45.7	60.5
14	0.1	0.2	0.48	0.049	0.142		0.059	0.96		48.3	63.0
15	0.11	0.2	0.54	0.054	0.055		0.52	1.02		47.0	66.0
20	0.12	0.79	0.51	0.049	0.158		0.51	1.01		45.7	61.0
21	0.11	0.26	0.46	0.053	0.057		0.058	1.5		45.7	51.8
23	0.11	0.21	0.46	0.052	0.055		0.49	1.44		48.3	65.0
19	0.11	0.29	0.5	0.05	0.053		0.054	2.04		38.1	55.9
24	0.1	0.29	0.48	0.053	0.054		0.5	1.95		39.4	56.9
22	0.12	0.23	0.52	0.051	0.056		0.056	2.47		29.2	60.5
25	0.11	0.23	0.44	0.05	0.055		0.48	2.44		34.3	39.6
Effect of Chromium and Copper Additions (0.2% Carbon Steel)											
8	0.2	0.22	0.52	0.053	0.057		0.52			63.5	70.6
10	0.23	0.22	0.55	0.053	0.049		0.056	0.96		57.2	69.1
16	0.21	0.23	0.54	0.052	0.054		0.51	0.96		50.8	73.7
Effect of Aluminum and Chromium Additions											
27	0.13	0.27	0.69	0.043	0.048		0.017			68.6	61.5
28	0.12	0.34	0.57	0.039	0.045		0.014	2.61		27.9	34.0
29	0.13	0.31	0.66	0.043	0.049		0.017	2.63	0.14 Al	30.5	31.5
26	0.11	0.17	0.45	0.05	0.046		0.075		1.59 Al	40.6	68.6
30	0.1	0.78	0.55	0.036	0.05		0.064	2.79	1.44 Al	24.1	24.9
Effect of Chromium, Copper, and Nickel Additions											
27	0.13	0.27	0.69	0.043	0.048		0.017			68.6	61.5
31	0.11	0.23	0.5	0.04	0.048	3.06	0.02			54.6	62.5
32	0.1	0.27	0.49	0.04	0.045	3.11	0.49			54.6	67.1
35	0.11	0.24	0.65	0.038	0.046	1.51	1.01			57.2	54.4
33	0.12	0.22	0.52	0.038	0.052	3.05	0.015	0.1		40.6	67.6
34	0.12	0.25	0.5	0.046	0.049	2.98	0.45	1.13		39.4	64.0
Effect of Chromium, Molybdenum, and Nickel Additions											
36	0.05	0.17	0.34	0.012	0.019		0.01	0.01		71.1	74.2
40	0.1	0.11	0.38	0.011	0.022		0.016		0.55 Mo	64.8	67.1
41	0.1	0.13	0.34	0.012	0.024	0.21	0.54		0.55 Mo	59.7	51.3
43	0.11	0.15	0.31	0.013	0.022	1.96	0.02	0.02	0.55 Mo	54.6	67.6
42	0.11	0.17	0.42	0.012	0.023		0.02	0.95	0.54 Mo	41.9	52.8
44	0.12	0.13	0.39	0.026	0.023	3	0.018	0.76	0.51 Mo	38.1	73.2
46	0.18	0.19	0.49	0.013	0.018	0.11	0.099	3.14	0.44 Mo	25.4	33.5
Effect of Sundry Additions											
47	0.06	0.13	0.47	0.04	0.015	0.09	0.23	0.04	0.05 W	59.7	63.5
48	0.09	0.12	0.47	0.037	0.013	0.1	0.21	0.03	0.05 W 0.295 Sb	62.2	54.9
49	0.08	0.13	0.36	0.039	0.014	0.1	0.74	0.06	0.06 W 0.465 As	58.4	70.1
60	0.1	0.11	0.37	0.037	0.014	0.1	0.22	0.05	0.06 W 0.41 Be	44.5	42.2
53	0.09	0.11	0.4	0.039	0.013	0.1	0.21	0.05	0.06 W 0.225 Bi	67.3	76.2
51	0.09	0.13	0.45	0.041	0.014	0.1	0.2	0.04	0.05 W <0.01 Cd	64.8	69.1
52	0.09	0.12	0.45	0.038	0.014	0.09	0.22	0.04	0.05 W 0.13 Pb	61.0	56.9
58	0.1	0.09	0.3	0.038	0.014	0.11	0.23	0.15	0.215 W 0.05 Ta 0.27 Nb	61.0	58.4
59	0.1	0.09	0.3	0.038	0.014	0.11	0.23	0.15	0.22 W 0.06 Ta 0.25 Nb	58.4	59.9
57	0.1	0.17	0.34	0.033	0.012	0.1	0.23	0.09	0.135 W 0.35 Ta 0.15 Nb	57.2	53.8
50	0.09	0.1	0.46	0.041	0.013	0.09	0.22	0.03	0.055 W 0.61 Sn	61.0	58.4
45	0.11	0.13	0.55	0.022	0.02	0.016	0.07		0.24 V	64.8	60.5
54	0.09	0.12	0.36	0.043	0.014	0.1	0.2	0.08	0.66 W	59.7	63.5
55	0.09	0.12	0.36	0.043	0.014	0.1	0.2	0.08	0.66 W	57.2	61.5

J.C. Hudson and J.F. Stanners, "The Corrosion Resistance of Low-Alloy Steels," in Journal of the Iron And Steel Institute, Volume 180, July 1955, pg. 271

Table III-H Compositions of the Materials Investigated by Hadfield and Main
5 Year Corrosion Study
(Hadfield and Main, 1936)

		Composition %							
Key	Material	C	Si	S	P	Mn	Cr	Ni	Cu
A	Medium C Steel, low S and P								
E	Medium C Steel, low S and P	0.34 - 0.35	0.0 - 0.2	0.024 - 0.026	0.0 - 0.027	0.71 - 0.72	----	----	----
F	Mild Steel, 0.7% Mn	0.23 - 0.25	0.0 - 0.16	0.061 - 0.064	0.0 - 0.048	0.0 - 0.68	----	----	----
B	Mild Steel, low Mn, high S and P	0.21 - 0.22	0.0 - 0.17	0.09 - 0.115	0.06 - 0.073	0.33 - 0.35	----	----	----
C	Mild Steel, 0.7% Mn	----	----	----	----	----	----	----	----
D	0.4% C Steel	0.4	0.2	0.049	0.045	0.85	----	----	----
G	Mild Steel, 0.5% Cu	0.21	0.14	0.043	0.046	0.94	----	----	0.63 - 0.64
H	Mild Steel, 2% Cu	.22 - .23	0.14	0.033 - 0.035	0.041 - 0.043	0.9 - 0.92	----	----	2.18 - 2.19
J	Cr Steel	0.36	0.22	----	----	0.13	13.57	----	----
S	Cr Steel	0.12	0.53	0.042	0.015	0.21	11.9	----	----
T	Cr Steel	0.12	0.53	0.042	0.015	0.21	11.9	----	----
U	Cr Steel	0.12	0.53	0.042	0.015	0.21	11.9	----	----
V	Cr Steel	0.12	0.53	0.042	0.015	0.21	11.9	----	----
W	Cr Steel	0.12	0.53	0.042	0.015	0.21	11.9	----	----
X	Cr Steel	0.12	0.53	0.042	0.015	0.21	11.9	----	----
K	3.5% Ni Steel	0.31	0.18	0.029	0.038	0.54	----	3.75	----
L	36% Ni Steel	0.09 - 0.15	0.0 - 0.09	----	----	0.77 - 0.97	----	36.45 - 36.67	----
M	Ingot Iron	0.03 - 0.04	0.03 - 0.04	0.032 - 0.033	0.016 - 0.019	0.06 - 0.07	----	----	----
N	Wrouht Iron	0.01 - 0.02	0.14 - 0.15	0.012 - 0.013	0.113 - 0.114	tr - .03	----	----	----
P	Swedish Charcoal Iron	0.02 - 0.04	0.0 - 0.03	0.0 - 0.019	0.0 - 0.022	0.01 - 0.03	----	----	----
Q	Hot-Blast Cast Iron	0.94 - 2.95	2.16 - 2.52	0.103 - 0.105	0.84 - 0.856	0.26 - 0.28	----	----	----
R	Cold-Blast Cast Iron	0.72 - 3.47	1.66	0.067	0.455	0.74	----	----	----

Reference:

R.A. Hadfield and S.A. Main, "Corrosion of Iron and Steel," in Journal of the Institution of Civil Engineers, Volume 3, 1935-1936, p. 3.

Table III-I Effect of Millscale on Corrosion of Carbon Steel

(Hadfield and Main, 1936)

			General (µm/ yr)		Deepest Pit (µm)	
Key	Material	Surface Cond.	Mean Tide	Immersion	Mean Tide	Immersion
				Halifax		
E	Medium C Steel, Low S and P	W/ millscale	69.2	95.0	5100.0	1860.0
A	Medium C Steel, Low S and P	W/O millscale	44.0	107.8	2520.0	760.0
F	Mild Steel, 0.7% Mn	W/ millscale	84.8	95.0	3100.0	1520.0
C	Mild Steel, 0.7% Mn	W/O millscale	49.4	93.2	1020.0	1400.0
				Ankland		
E	Medium C Steel, Low S and P	W/ millscale	30.2	87.4	590.0	2790.0
A	Medium C Steel, Low S and P	W/O millscale	54.8	116.8	550.0	620.0
F	Mild Steel, 0.7% Mn	W/ millscale	31.6	89.0	390.0	2320.0
C	Mild Steel, 0.7% Mn	W/O millscale	59.2	99.0	nil	440.0
				Plymouth		
E	Medium C Steel, Low S and P	W/ millscale	61.0	93.0	1730.0	1720.0
A	Medium C Steel, Low S and P	W/O millscale	114.8	106.0	1130.0	1340.0
F	Mild Steel, 0.7% Mn	W/ millscale	69.2	105.2	1220.0	3130.0
C	Mild Steel, 0.7% Mn	W/O millscale	117.0	128.4	990.0	690.0
				Colombo		
E	Medium C Steel, Low S and P	W/ millscale	330.0	117.8	4560.0	2760.0
A	Medium C Steel, Low S and P	W/O millscale	323.4	145.0	1970.0	1170.0
F	Mild Steel, 0.7% Mn	W/ millscale	230.6	100.8	5300.0	3150.0
C	Mild Steel, 0.7% Mn	W/O millscale	197.6	112.6	740.0	1030.0

		General (µm/ yr)		Deepest Pit (µm)	
Key	Material	Mean Tide	Immersion	Mean Tide	Immersion
		Halifax			
N	Wrought	44.2	108.0	2581.0	1010.0
P	Swedish Charcoal	67.6	108.2	4090.0	nil
M	Ingot	78.0	114.0	2940.0	600.0
D	0.4% Carbon Steel	82.8	92.8	4720.0	960.0
F	Mild Steel, 0.7% Mn	84.8	95.0	3100.0	1520.0
B	Mild Steel, Low Mn, High S and I	133.8	98.2	1220.0	1810.0
E	Medium C Steel, Low S and P	69.2	95.0	5100.0	1860.0
R	Cold-Blast Cast Iron	4.6	55.8		
Q	Hot-Blast Cast Iron	7.0	64.8		
H	Mild Steel, 2% Cu	41.6	90.0	4710.0	790.0
G	Mild Steel, 0.5% Cu	37.0	88.2	4180.0	2250.0
L	36% Ni	10.0	68.2	800.0	1900.0
K	3.5% Ni	12.4	86.0	1080.0	1740.0
J	13% Cr Steel	12.8	55.4	1180.0	3200.0
		Ankland			
N	Wrought	35.8	80.2	nil	1520.0
P	Swedish Charcoal	47.6	80.4	nil	1290.0
M	Ingot	54.4	85.0	200.0	2190.0
D	0.4% Carbon Steel	26.6	96.6	810.0	2280.0
F	Mild Steel, 0.7% Mn	31.6	89.0	590.0	2790.0
B	Mild Steel, Low Mn, High S and I	26.2	96.6	690.0	3680.0
E	Medium C Steel, Low S and P	30.2	87.4	390.0	2320.0
R	Cold-Blast Cast Iron	8.2	58.4		
Q	Hot-Blast Cast Iron	14.0	64.2		
H	Mild Steel, 2% Cu	26.4	79.0	740.0	3190.0
G	Mild Steel, 0.5% Cu	16.2	81.4	720.0	2500.0
L	36% Ni	6.2	35.4	nil	1190.0
K	3.5% Ni	18.2	66.2	950.0	2710.0
J	13% Cr Steel	12.8	53.8	630.0	1400.0
		Plymouth			
N	Wrought	85.2	81.8	740.0	920.0
P	Swedish Charcoal	99.8	101.4	990.0	750.0
M	Ingot	104.2	108.2	1090.0	1880.0
D	0.4% Carbon Steel	75.8	94.4	2730.0	2050.0
F	Mild Steel, 0.7% Mn	69.2	105.2	1730.0	1720.0
B	Mild Steel, Low Mn, High S and I	77.0	83.2	1040.0	2140.0
E	Medium C Steel, Low S and P	61.0	93.0	1730.0	3130.0
R	Cold-Blast Cast Iron	29.8	34.8		
Q	Hot-Blast Cast Iron	40.8	36.4		
H	Mild Steel, 2% Cu	58.2	85.0	1930.0	1730.0
G	Mild Steel, 0.5% Cu	74.2	82.2	2250.0	2820.0
L	36% Ni	22.8	51.8	nil	1280.0
K	3.5% Ni	52.0	71.0	1620.0	3230.0
J	13% Cr Steel	28.8	31.4	670.0	830.0
		Colombo			
	Wrought	161.8	90.8	1710.0	1850.0
P	Swedish Charcoal	166.4	77.6	nil	4850.0
M	Ingot	173.6	98.2	2000.0	5030.0
D	0.4% Carbon Steel	212.6	116.2	5630.0	2450.0
F	Mild Steel, 0.7% Mn	230.6	100.8	4560.0	2760.0
B	Mild Steel, Low Mn, High S and I	181.0	102.4	2300.0	6000.0
E	Medium C Steel, Low S and P	330.0	117.8	5300.0	3150.0
R	Cold-Blast Cast Iron	237.6	110.6		
Q	Hot-Blast Cast Iron	259.8	115.0		
H	Mild Steel, 2% Cu	242.2	100.8	4410.0	3330.0
G	Mild Steel, 0.5% Cu	216.4	84.6	5090.0	3010.0
L	36% Ni	42.6	29.8	1730.0	1380.0
K	3.5% Ni	167.0	77.2	na	4420.0
J	13% Cr Steel	81.4	57.6	6500.0	6500.0

Table III-J Corrosion Behavior of Carbon Steel, Nickel Steel and Chromium Steel

(Hadfield & Main, 1936)

	General Corrosion (μm/yr)		Pitting (μm)	
	Mean Tide	Immersion	Mean Tide	Immersion
Halifax				
0.4% C steel	82.8	92.8	4720	960
36% Ni steel	10.0	68.2	800	1900
13% Cr steel	12.8	55.4	1180	3200
Auckland				
0.4% C steel	266.0	96.6	810	2280
36% Ni steel	6.2	35.4	nil	1190
13% Cr steel	12.8	53.8	630	1400
Plymouth				
0.4% C steel	75.8	94.4	2730	2050
36% Ni steel	22.8	51.8	nil	1280
13% Cr steel	28.8	31.4	670	830
Colombo				
0.4% C steel	212.6	116.2	5630	2450
36% Ni steel	42.6	29.8	1730	1380
13% Cr steel	81.4	57.6	6500	6500

Table III-K Corrosion in Deep Ocean Environments

(Wheatfall, 1967; Reinhart, 1965, and Fink & Boyd, 1970)

	Steel	C	Mn	P	S	Si	Cr	Ni	Mo	Cu	V	Other	Depth (ft)	Duration (yr)	Location	Corrosion Rate (µm/yr)	Comments	Reference
1	AISI 1015	----	----	----	----	----	----	----	----	----	----	----	5640	0.34	Port Hueme, Ca	77	Crevice -- 178 µm	Wheatfall, 1967.
2	AISI 1015	----	----	----	----	----	----	----	----	----	----	----	5640	2.1	Pacific	44	Crevice -- 254 µm	Wheatfall, 1967.
3	AISI 1015	----	----	----	----	----	----	----	----	----	----	----	5300	2.9	Ocean	15	Crevice -- 508 µm	Wheatfall, 1967.
4	AISI 1015	----	----	----	----	----	----	----	----	----	----	----	surface	1.1	"	133	Crevice -- 381 µm	Wheatfall, 1967.
	HSLA	0.18	0.86	0.014	0.023	0.28	0.64	0.05	0.18	----	----	----	5640	0.34	"	73.2	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	5640 (mud)	0.34	"	55.6	Slight Crevice	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	5300	2.9	"	15.2	Slight Crevice	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	5300 (mud)	2.9	"	18.8	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	2340	0.54	"	35.6	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	2340 (mud)	0.54	"	35.1	Uniform	Reinhart, 1965.
	HSLA	0.12	0.3	0.015	0.025	0.27	1.25	2.34	0.2	0.17	0.047	0.02 Ti, 0.0028 B	5640	0.34	"	118.6	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	5640 (mud)	0.34	"	105.4	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	5300	2.9	"	13.5	Slight Crevice	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	2340	0.54	"	36.6	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	2340 (mud)	0.54	"	36.1	Uniform	Reinhart, 1965.
	HSLA	0.17	0.28	0.02	0.018	0.2	1.76	2.96	0.4	----	----	----	5300	2.9	"	17	Slight Crevice	Reinhart, 1965.
	AISI 4340	----	----	----	----	----	----	----	----	----	----	----	5640	0.34	"	70.4	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	2340	0.54	"	36.1	Uniform	Reinhart, 1965.
	"	----	----	----	----	----	----	----	----	----	----	----	2340 (mud)	0.54	"	36.6	Uniform	Reinhart, 1965.
	Carbon Steel	----	----	----	----	----	----	----	----	----	----	----	Surface	91	Atlantic	201	----	Fink & Boyd, 1970.
	"	----	----	----	----	----	----	----	----	----	----	----	"	167	Ocean	127	----	Fink & Boyd, 1970.
	"	----	----	----	----	----	----	----	----	----	----	----	"	365	"	108	----	Fink & Boyd, 1970.
	"	----	----	----	----	----	----	----	----	----	----	----	"	1200	"	76.2	----	Fink & Boyd, 1970.
	1010 C Steel	0.11 -	0.5 -	0.004 -	0.023 -	0.048 -	----	----	----	----	----	----	5600	0.3	Atlantic	92.5	----	Fink & Boyd, 1970.
	"	0.12	0.52	0.016	0.024	0.6	----	----	----	----	----	----		2.9	Ocean	45.3	----	Fink & Boyd, 1970.
	Wrought Iron	0.02	0.06	0.13	0.01	0.13	----	----	----	----	----	----	5300	2.9	Fort Hueme, Ca	15.2	----	Fink & Boyd, 1970.
	"	----	----	----	----	----	----	----	----	----	----	----	2340	0.54	Pacific Ocean	50.8	----	Fink & Boyd, 1970.
	ASTM A 36	0.2 -	0.55 -	0.01 -	0.02 -	0.055	----	----	----	----	----	----	5300	2.9	"	15.2	----	Fink & Boyd, 1970.
	"	0.24	0.7	0.011	0.027	0.064	----	----	----	----	----	----	2340	0.54	"	43.2	43.2 µm crevice	Fink & Boyd, 1970.
	HSLA	0.07	0.28	0.014	0.01	0.025	0.88	0.31	----	0.28	----	----	5300	2.9	"	27.9	17.8 µm crevice	Fink & Boyd, 1970.
	HSLA	0.14	0.38	0.11	0.025	0.54	0.56	0.74	0.42	0.22	0.36	.0041 B	5300	2.9	"	17.8	20.3 µm crevice	Fink & Boyd, 1970.
	"												2340	0.54	"	35.6	35.6 µm crevice	Fink & Boyd, 1970.

Table III-L Corrosion of Steels in River Waters

(ASM, 1978; Southwell & Alexander, 1970; and Larrabee, 1953)

		Composition (%)							
Key	Metal	C	Mn	P	S	Si	Cr	Ni	Cu
	Tests in Monongahela, Allegheny and Mississippi Rivers								
HSLA	Cor-Ten	0.08	0.34	0.1	0.026	0.45	0.88	0.53	0.37
HSLA	Tri-Ten	0.19	0.92	0.015	0.028	0.18	0.05	0.9	0.36
Copper	Man-Ten	0.21	1.37	0.02	0.024	0.21	–	–	0.28
Copper	Structural Copper	0.2	0.45	0.017	0.026	0.046	0.06	0.11	0.34
Plain C	Structural Silicon	0.39	0.78	0.011	0.028	0.22	0.02	0.03	0.042
Plain C	Structural Carbon	0.19	0.4	0.009	0.034	0.032	0.035	0.018	0.028
	Tests in Gatineau River, Canada								
	Plain C	0.07 to 0.09	0.3 to 0.6						0.03 to 0.06
	Copper	0.07 to 0.09	0.3 to 0.6						0.2 to 0.28
	HSLA	0.07 to 0.1	0.65 to 0.75	0.06 to 0.08		0.05 to 0.1		0.55 to 0.65	1.05 to 1.15

		1 Year Exposure*		2.5 Year Exposure*		4.5 Year Exposure*		8 Year Exposure**	
Key	Steel	Average (μM/Yr)	Max Pit (μM)	Average (μM/Yr)	Max Pit (μM)	Average (μM/Yr)	Max Pit (μM)	Average (μM/Yr)	Max Pit (μM)
		Monongalela River, Charleroi, Pa – pH = 3.5 - 4 Ferric Sulfate Contaminated							
HSLA	Cor-Ten	124.5	203.2	78.7	304.8	68.6	355.6	45.4	700
HSLA	Tri-Ten	177.8	330.2	116.8	457.2	101.6	533.4	45.4	700
Copper	Man-Ten	167.6	406.4	96.5	482.6	76.2	431.8	61.2	965
Copper	Structural Copper	162.6	304.8	104.1	482.6	88.9	736.6	61.2	965
Plain C	Structural Silicon	162.6	330.2	132.1	508.0	88.9	787.4	73.3	990
Plain C	Structural Carbon	165.1	584.2	116.8	711.2	104.1	533.4	73.3	990
		Allegheny River, Kittanig, Pa - Low Degree of Turbulence – Oxygen Concentration Cells							
HSLA	Cor-Ten	88.9	584.2	88.9	939.8	61.0	889.0	55	1237
HSLA	Tri-Ten	96.5	1066.8	91.4	889.0	66.0	812.8	55	1237
Copper	Man-Ten	101.6	736.6	88.9	1066.8	76.2	1219.2	59	1097
Copper	Structural Copper	106.7	609.6	91.4	787.4	63.5	863.6	59	2097
Plain C	Structural Silicon	106.7	584.2	94.0	965.2	73.7	1041.4	61	1467
Plain C	Structural Carbon	106.7	660.4	94.0	939.8	71.1	1498.6	61	1467
		Mississippi River, Winfield Mo. – Stagnant Water							
HSLA	Cor-Ten	35.6	152.4	30.5	330.2	25.4	762.0	28	1640
HSLA	Tri-Ten	40.6	279.4	35.6	431.8	30.5	711.2	28	1640
Copper	Man-Ten	38.1	304.8	33.0	355.6	27.9	635.0	34	1270
Copper	Structural Copper	38.1	254.0	33.0	254.0	25.4	482.6	34	1270
Plain C	Structural Silicon	40.6	203.2	35.6	330.2	27.9	431.8	32	1393
Plain C	Structural Carbon	38.1	279.4	35.6	355.6	27.9	787.4	32	1393
		Gatineau River, Canada – Unpolluted							
	Plain Carbon			47					
	Copper			37					
	HSLA			32					

References:

*Monongahela, Allegheny and Mississippi – 1, 2.5, and 4.5 year tests

C.P. Larrabee, "Corrosion Resistance of High-Strength Low-Alloy Steels as Influenced by Composition and Environment," in Corrosion Vol 9, #8, p 259

**Monongahela, Allegheny and Mississippi – 8 yr and Gatineau

Metals Handbook, 9th Edition, Volume. 1, page 733, American Society for Metals, 1978.

Table III-M Corrosion Performance Comparison Between Plain Carbon Steel and Alloyed Steels in Four Aqueous Environments

(ASM, 1978; Southwell & Alexander, 1970; and Larrabee, 1953)

Location	Exposure Time (yr)	Plain Carbon Steel (µm/yr)	Alloyed Steels (µm/yr)
Panama	4	108	82-101
Monongahela	4.5	104	69-100
Allegheny	4.5	71	61-76
Mississippi	4.5	28	25-31

Table III-N Corrosion of Metal Alloys in Utah Geothermal Waters

(Hong & Pitt, 1987)

Steel	C	Si	Mn	S	P	Ni	Cr	Other
9Cr-1Mo	0.089	0.09	0.4	0.003	0.003	0.09	9.03	0.95Mo, 0.2V, 0.17Nb
A-36	0.250	0.20	1.0	0.050	0.040	--	--	0.2 Cu

Reference:
Y.K. Hong and C.H. Pitt, "Corrosion of Selected Metal Alloys in Utah Geothermal Waters," in Journal of Materials for Energy Systems, Vol. 5, No. 2, Sept. 1983, P. 77

Temperature Degree C	Corrosion Rate (μm/yr)	Pit Density (#/Cm^3)	Maximum Pit Depth (μm)
	9Cr - 1 Mo Alloy Steel		
25	25.4	43	92
50	218.4	268	680
60	246.4	85	1200
75	157.5	50	800
	A-36		
25	40.6	Severe pitting occurred	
50	50.8	at temp = 25 degrees C	
90.9	287.0		
105.6	353.0		

Table III-O Corrosion of Mild Steels In a Granitic Environment

(Marsh & Taylor, 1988)

Steel	C	Si	Mn	S	P
Forged	0.2	0.08	0.6	0.038	0.01
Cast	0.23	0.42	0.89	0.005	0.011

90 degree C Tests

Steel	Backfill/ Depth (cm)	Test Time (yr)	Avg. Corrosion Rate (μm/yr)
Forged	Granite / 5	0.5	19.2
Cast	Granite / 5	0.5	22.8
Forged	Bentonite / 5	1.3	7.7
Cast	Bentonite / 5	1.3	1.7
Forged	Granite / 10	2.55	24.5
Cast	Granite / 10	2.55	26.0
Forged	Bentonite / 10	2.28	7.1
Cast	Bentonite /10	2.28	2.6

Reference:

G.P. Marsh and K.J. Taylor, "An Assessment of Carbon Steel Containers For Radioactive Waste Disposal," in Corrosion Science, Vol. 28, No. 3, 1988, p. 289

Table III-P Corrosion of Carbon Steels Under Tuff Repository Conditions, 62.5 Days (0.17 yr)

(McCright & Weiss, 1985)

Steel	C	Si	Mn	S	P	Mo	Cr
Gray Cast Iron	3.26	2.68	0.72	0.058	0.038	--	--
A-36	0.22	0.03	0.46	0.009	0.008	--	--
1020	0.18	--	0.35	0.025	0.04	--	--
2.25Cr-1Mo	0.12	0.27	0.56	0.012	0.016	0.94	2.36
9-Cr-1Mo	0.13	0.71	0.48	0.01	--	0.96	8.27

Temperature Degree C	General Corrosion Rate (µm/yr)	Pitting Corrosion Rate (µm/yr)	Crevice Corrosion Rate (µm/yr)
	1020 Carbon Steel		
50	401.0	380.0	413.0
70	505.0	1018.0	359.0
80	531.0	465.0	472.0
90	414.0	1046.0	563.0
100	320.0	1018.0	635.0
	Gray Cast Iron		
50	359.0	203.0	98.2
70	422.0	392.0	138.0
80	357.0	79.6	305.0
90	323.0	27.9	14.6
100	318.0	0.0	232.0
	2.25 Cr - 1 Mo Alloy Steel		
50	316.0	649.0	946.0
70	469.0	1448.0	1100.0
80	376.0	1089.0	1415.0
90	370.0	868.0	800.0
100	278.0	1352.0	781.0
	9Cr - 1Mo Alloy Steel		
50	21.2	42.3	202.0
70	21.2	152.0	450.0
80	14.1	246.0	319.0
90	8.3	0.0	84.7
100	6.8	0.0	33.0

Reference:
R.D. McCright and H. Weiss, "Corrosion Behavior of Carbon Steels Under Tuff Repository Conditions,"
in Material Research Society Symposium Proceedings, Vol. 44, 1985
p. 287

Table IV-A Corrosion of Cast and Wrought Ferrous Materials in Select Soils with Differing Aeration

(Romanoff, 1957)

Metal ID		Material	Exposure (years)	Composition, % C	Si	Mn	S	P	Cr	Ni	Cu	Mo	Other
		Cast Ferrous											
G	1932	Rattled Cast Iron	14.3	3.58	1.64	0.48	0.07	0.79	–	–	–	–	–
A	1941	Plain Cast Iron	11.0	3.22	2.19	0.91	0.12	–	–	–	–	–	–
I	1932	Low Alloy (0.51Cu-0.18Mn)	14.3	2.53	1.43	0.28	0.08	0.13	–	–	0.51	–	–
J	1932	Low Alloy (0.62Cu-0.83Mn)	14.3	2.90	2.04	0.83	0.06	0.25	–	–	0.62	–	–
C	1932	Low Alloy (0.30Cr-0.15Ni)	14.3	3.50	2.50	0.70	0.05	0.40	0.30	0.15	–	–	–
B	1941	Low Alloy (1.27Ni-0.32Cu)	11.0	3.28	2.09	0.83	0.12	–	–	1.27	0.32	–	–
C	1941	Low Alloy (1.71Ni-0.98Cu)	11.0	3.24	2.08	0.80	0.12	–	–	1.71	0.98	–	–
NC	1941	Low Alloy (2.08Ni-1.10Cu)	11.0	2.80	2.03	–	–	–	–	2.08	1.10	–	–
N	1941	Low Alloy (3.1Ni)	11.0	2.75	2.00	–	–	–	–	3.10	–	–	–
D	1941	Low Alloy (3.32Ni)	11.0	3.21	2.11	0.72	0.12	–	–	3.32	–	–	–
E	1932	High Alloy (15.0Ni-6.6Cu-2.6Cr)	14.3	2.98	2.13	1.00	–	–	2.61	15.00	6.58	–	–
		Wrought Ferrous											
H	1932	Open Hearth Iron (0.62Cu-0.15Mo)	14.3	0.04	0.05	0.32	0.027	0.016	–	–	0.52	0.15	–
O	1937	Open Hearth Iron (0.45Cu-0.07Mo)	12.7	0.03	0.003	0.16	0.032	0.007	0.02	0.15	0.45	0.07	–
N	1937	Open Hearth Iron (0.54Cu-0.13Mo)	12.7	0.06	0.001	0.098	0.029	0.069	0.02	0.14	0.54	0.13	–
MM	1939	Open Hearth Iron (0.51Cu-0.07Mo)	12.1	0.04	–	0.16	0.027	0.008	0.04	0.14	0.51	0.057	0.002Sn
H	1941	Open Hearth Iron (0.47Cu-0.08Mo)	11.0	0.036	0.005	0.16	0.025	0.008	–	–	0.47	0.08	–
J	1937	Cu-Ni Steel (0.95Cu-0.52Ni)	12.7	0.06	0.047	0.49	0.025	0.095	–	0.52	0.95	–	–
B	1937	Ni-Cu Steel (1.96Ni-1.01Cu)	12.7	0.07	0.14	0.44	0.02	0.01	–	1.96	1.01	–	–
D	1932	Ni-Cu Steel (2.47Ni-1.08Cu)	14.3	0.14	0.19	0.21	–	–	–	2.47	1.08	–	–
Y	1941	Ni-Cu Steel (1.95Ni-1.04Cu)	11.0	0.12	0.12	0.58	0.025	0.07	–	1.95	1.04	–	–
C	1937	Cr-Si-Cu Steel (1.02Cr-0.42Cu)	12.7	0.075	0.84	0.2	0.018	0.124	1.02	0.22	0.428	–	–
NN	1939	2Cr Steel w/ Mo	12.1	0.09	0.25	0.46	0.01	0.015	1.96	–	–	0.49	–
KK	1937	2.01Cr-0.57Mo Steel	12.7	0.082	0.51	0.46	0.015	0.017	2.01	0.07	0.004	0.57	–
P	1932	5.05Cr Steel	14.3	0.13	–	0.46	0.025	0.012	5.05	–	–	–	–
D	1937	5.02Cr Steel	12.7	0.077	0.43	0.37	0.005	0.015	5.02	0.09	0.008	–	–
E	1937	4.67Cr-0.51Mo Steel	12.7	0.074	0.41	0.32	0.006	0.013	4.67	0.09	0.004	0.51	0.03Al-0.022Ti
H	1937	5.76Cr-0.43Mo Steel	12.7	0.06	0.39	0.4	0.014	0.021	5.76	0.17	0.004	0.43	0.27Al

Table IV-A Corrosion of Cast and Wrought Ferrous Materials in Select Soils with Differing Aeration

(Romanoff, 1957) -- (Continued)

	Good Aeration							
	Average Penetration, µm/yr				Maximum Penetration, µm			
Material	Cecil clay Loam	Hagers-town Loam	Susque-hanna Clay	Chino Silt Loam	Cecil clay Loam	Hagers-town Loam	Susque-hanna Clay	Chino Silt Loam
Cast Ferrous								
Rattled Cast Iron	8.5	6.2	21.8	23.1	3505	3785	3556	3708
Plain Cast Iron	19.7	11.7	18.7	11.4	3683	3353	2972	2819
Low Alloy (0.51Cu-0.18Mn)	5.9	5.9	17.2	31.6	2337	3353	2616	3581
Low Alloy (0.62Cu-0.83Mn)	8.7	6.7	19.3	32.1	1194	2743	2845	3810
Low Alloy (0.30Cr-0.15Ni)	9.0	8.5	17.0	45.0	4343+	4674	3150	4902+
Low Alloy (1.27Ni-0.32Cu)	15.7	12.7	21.7	10.0	3454	4013	5055+	2896
Low Alloy (1.71Ni-0.98Cu)	18.4	13.7	22.4	8.7	3353	2896	4572+	2845
Low Alloy (2.08Ni-1.10Cu)	10.7	12.7	16.4	24.7	3785	3734	3861	3912
Low Alloy (3.1Ni)	12.4	13.7	17.4	28.7	4140	4343	3658	4013
Low Alloy (3.32Ni)	17.0	14.4	18.0	12.4	3048	3988	5664+	4699+
High Alloy (15.0Ni-6.6Cu-2.6Cr)	3.1	2.8	5.9	8.0	1118	1295	1524	1499
Wrought Ferrous								
Open Hearth Iron (0.62Cu-0.15Mo)	10.8	8.2	15.2	38.3	2667+	2261	2134	2642+
Open Hearth Iron (0.45Cu-0.07Mo)	10.7	9.5	15.0	25.7	1626	1930	1295	2946
Open Hearth Iron (0.54Cu-0.13Mo)	11.0	9.8	15.3	23.4	1499	1702	1727	1118
Open Hearth Iron (0.51Cu-0.07Mo)	12.1	8.5	18.5	26.1	2438	1829	1626	2692
Open Hearth Iron (0.47Cu-0.08Mo)	12.7	10.4	18.0	28.1	1981	1676	1219	3353+
Cu-Ni Steel (0.95Cu-0.52Ni)	8.7	7.5	15.9	25.5	1727	1829	2362	2007
Ni-Cu Steel (1.96Ni-1.01Cu)	9.3	9.0	23.7	31.2	2388	1880	2921	3150
Ni-Cu Steel (2.47Ni-1.08Cu)	8.7	6.2	9.2	17.7	1067	1803	1727	1854
Ni-Cu Steel (1.95Ni-1.04Cu)	12.4	9.0	16.7	14.4	2032	1549	1499	2413
Cr-Si-Cu Steel (1.02Cr-0.42Cu)	10.4	9.5	15.9	33.0	1981	2337	2235	2667
2Cr Steel w/ Mo	11.8	7.3	17.6	29.5	3353+	3023	2896	3683+
2.01Cr-0.57Mo Steel	10.1	7.2	13.6	26.3	1219	2896	1829	1778
5.05Cr Steel	7.5	5.9	15.7	30.6	1676	2159	3200	2769+
5.02Cr Steel	8.1	4.6	7.2	10.1	2134	1422	2870	1676
4.67Cr-0.51Mo Steel	8.4	4.9	6.9	11.0	2489	1372	1880	1778
5.76Cr-0.43Mo Steel	7.8	5.2	6.1	10.4	2083	1372	1778	1727

2 Specimens

KEY: Average Penetration

+ One Specimen Destroyed By Corrosion
D Both Specimens Destroyed By Corrosion

Maximum Penetration

+ ONE OR MORE SPECIMENS CONTAINED HOLES
D Both Specimens Destroyed By Corrosion

Table IV-A Corrosion of Cast and Wrought Ferrous Materials in Select Soils with Differing Aeration

(Romanoff, 1957) -- (Continued)

Fair Aeration

Material	Average Penetration, µm/yr			Maximum Penetration, µm		
	Docas Clay	Mojave Fine Gravelly Loam	Merced Silt Loam	Docas Clay	Mojave Fine Gravelly Loam	Merced Silt Loam
Cast Ferrous						
Rattled Cast Iron	D	D	--	6350+	6350+	--
Plain Cast Iron	83.8	17.7	--	5436+	4064+	6350+
Low Alloy (0.51Cu-0.18Mn)	127.4	47.8	--	6629+	8890+	--
Low Alloy (0.62Cu-0.83Mn)	148.3	58.1	--	6502+	8890+	--
Low Alloy (0.30Cr-0.15Ni)	92.8	D	--	5740+	6350+	--
Low Alloy (1.27Ni-0.32Cu)	49.4	22.7	D	5410+	3607	6350+
Low Alloy (1.71Ni-0.98Cu)	79.8	31.7	D	5232+	3632	6350+
Low Alloy (2.08Ni-1.10Cu)	88.5	30.1	124.3	4521	3886	6807+
Low Alloy (3.1Ni)	83.5	28.7	127.3	4953	3734	6731+
Low Alloy (3.32Ni)	70.5	31.4	D	6350+	2921	6350+
High Alloy (15.0Ni-6.6Cu-2.6Cr)	12.3	19.5	--	864	1372	--
Wrought Ferrous						
Open Hearth Iron (0.62Cu-0.15Mo)	80.9	D	--	3683+	3683+	--
Open Hearth Iron (0.45Cu-0.07Mo)	53.5	44.0	63.1	3632	3683	5690+
Open Hearth Iron (0.54Cu-0.13Mo)	52.1	43.7	57.9	3277	3581	4546+
Open Hearth Iron (0.51Cu-0.07Mo)	55.6	51.9	0.0	3708	4089	6350+
Open Hearth Iron (0.47Cu-0.08Mo)	95.2	74.8	73.1	3683+	3683+	3683+
Cu-Ni Steel (0.95Cu-0.52Ni)	67.1	73.8	57.9	3353	4369	4877
Ni-Cu Steel (1.96Ni-1.01Cu)	41.7	57.3	85.6	2591	2769	4039
Ni-Cu Steel (2.47Ni-1.08Cu)	65.5	37.5	--	3683+	3683+	--
Ni-Cu Steel (1.95Ni-1.04Cu)	65.5	D	D	4953	5334+	5334+
Cr-Si-Cu Steel (1.02Cr-0.42Cu)	22.0	50.9	D	1575	3251+	4575+
2Cr Steel w/ Mo	31.0	51.9	D	2845	3683+	3683+
2.01Cr-0.57Mo Steel	25.5	40.5	59.6	1676	3708+	4445+
5.05Cr Steel	D	D	--	3912+	--	3912+
5.02Cr Steel	20.8	42.5	66.0	3200	4013	1727
4.67Cr-0.51Mo Steel	18.5	42.2	52.4	3150+	4775+	2235
5.76Cr-0.43Mo Steel	24.6	44.0	89.7	3861+	5156+	1524

2 Specimens

KEY: Average Penetration

+ One Specimen Destroyed By Corrosion
D Both Specimens Destroyed By Corrosion

Maximum Penetration

+ ONE OR MORE SPECIMENS CONTAINED HOLES
D Both Specimens Destroyed By Corrosion

Table IV-A Corrosion of Cast and Wrought Ferrous Materials in Select Soils with Differing Aeration

(Romanoff, 1957) -- (Continued)

Poor Aeration

Material	Average Penetration, μm/yr			Maximum Penetration, μm		
	Muck	Rifle Peat	Sharkley Clay	Muck	Rifle Peat	Sharkley Clay
Cast Ferrous						
Rattled Cast Iron	D	73.5	31.6	6350+	5004+	2489
Plain Cast Iron	18	33.7	13.4	3835+	1803	2159
Low Alloy (0.51Cu-0.18Mn)	95.3	86.3	28.3	7569+	5537	2997
Low Alloy (0.62Cu-0.83Mn)	101.7+	78.9	27.5	7976+	3861	3150
Low Alloy (0.30Cr-0.15Ni)	D	65.0	27.7	6350+	6350+	3480
Low Alloy (1.27Ni-0.32Cu)	14.4	56.1	12.4	4826+	1549	4369
Low Alloy (1.71Ni-0.98Cu)	15	20.7	11.4	4267+	1651	3099
Low Alloy (2.08Ni-1.10Cu)	17.7	26.7	10.4	5232	2057	3531
Low Alloy (3.1Ni)	22.4	32.4	9.7	3632	2007	2642
Low Alloy (3.32Ni)	12.4	24.4	11.7	3454	1422	3353
High Alloy (15.0Ni-6.6Cu-2.6Cr)	25.7	29.8	13.4	1372	1829	1473
Wrought Ferrous						
Open Hearth Iron (0.62Cu-0.15Mo)	44.4	65.5	25.2	2997+	2692+	2642+
Open Hearth Iron (0.45Cu-0.07Mo)	50.6	51.8	24.9	2896	1270	1626
Open Hearth Iron (0.54Cu-0.13Mo)	49.2	60.2	24.9	1803	1118	2235
Open Hearth Iron (0.51Cu-0.07Mo)	51	55.9	27.6	3073	1575	2210
Open Hearth Iron (0.47Cu-0.08Mo)	34.7	63.1	24.0	2438	1499	1981
Cu-Ni Steel (0.95Cu-0.52Ni)	54.1	41.4	22.9	4394	1524	2540
Ni-Cu Steel (1.96Ni-1.01Cu)	60.2	60.8	22.0	2489	1600	4572
Ni-Cu Steel (2.47Ni-1.08Cu)	33.7	44.7	15.9	2692+	2337	1067
Ni-Cu Steel (1.95Ni-1.04Cu)	37.1	53.8	21.7	2769	2235	2388
Cr-Si-Cu Steel (1.02Cr-0.42Cu)	54.1	52.1	20.5	2692	2718	1422
2Cr Steel w/ Mo	49.2	58.0	22.5	2870	1930	2210
2.01Cr-0.57Mo Steel	40.5	44.3	19.1	1880	1803	1651
5.05Cr Steel	52.9	70.4	23.4	3912+	3912+	2388
5.02Cr Steel	22.6	49.8	5.5	1727	1753	1067
4.67Cr-0.51Mo Steel	27.5	43.4	7.5	2235	1880	965
5.76Cr-0.43Mo Steel	20.5	41.7	5.2	1524	2540	1118

2 Specimens

KEY: Average Penetration

+ One Specimen Destroyed By Corrosion
D Both Specimens Destroyed By Corrosion

Maximum Penetration

+ ONE OR MORE SPECIMENS CONTAINED HOLES
D Both Specimens Destroyed By Corrosion

Table IV-A Corrosion of Cast and Wrought Ferrous Materials in Select Soils with Differing Aeration

(Romanoff, 1957) -- (Continued)

	Very Poor Aeration									
	Average Penetration, µm/yr					Maximum Penetration, µm				
Material	Acadia Clay	Lake Charles Clay	Carlisle Muck	Tidal Marsh	Cinders	Acadia Clay	Lake Charles Clay	Carlisle Muck	Tidal Marsh	Cinders
Cast Ferrous										
Rattled Cast Iron	D	D	17.5	33.7	D	6350+	6350+	1778	3962	6350+
Plain Cast Iron	–	64.5	53.4	14.4	D	–	6350+	2718	2337	6350+
Low Alloy (0.51Cu-0.18Mn)	70.9	107.9	13.1	19.0	D	8052	5334	2362	3302	8890+
Low Alloy (0.62Cu-0.83Mn)	79.9+	110.5	15.9	21.8	132.3+	8890	5842	1448	1981	7976+
Low Alloy (0.30Cr-0.15Ni)	D	87.4+	12.8	11.6	D	6350+	5634+	1422	3658	6350+
Low Alloy (1.27Ni-0.32Cu)	–	72.5	51.8	16.0	D	–	6350+	2718	3861+	6350+
Low Alloy (1.71Ni-0.98Cu)	–	67.8	52.1	12.0	D	–	5410+	2616	3150+	6350+
Low Alloy (2.08Ni-1.10Cu)	–	75.5	53.4	26.7	D	–	5486	2870	1397	6350+
Low Alloy (3.1Ni)	–	90.9	54.8	21.4	D	–	6299	3251	4750	6350+
Low Alloy (3.32Ni)	–	64.8	51.4	6.7	D	–	4877+	2692	5740	6350+
High Alloy (15.0Ni-6.6Cu-2.6Cr)	24.9	44.4	3.3	3.9	D	1981	1829	762	762	6350+
Wrought Ferrous										
Open Hearth Iron (0.62Cu-0.15Mo)	56.3	59.1	11.0	20.6	D	2667+	2743+	965	1880	3683+
Open Hearth Iron (0.45Cu-0.07Mo)	–	D	29.2	46.0	88.2	0	6172+	1473	1524	3556
Open Hearth Iron (0.54Cu-0.13Mo)	–	D	21.4	33.8	85.3	0	6350+	1372	1981	3861
Open Hearth Iron (0.51Cu-0.07Mo)	–	D	24.0	37.7	D	0	6350+	2083	2184	6350+
Open Hearth Iron (0.47Cu-0.08Mo)	–	D	25.7	19.0	D	0	3683+	2057	2769	3683+
Cu-Ni Steel (0.95Cu-0.52Ni)	–	155.9	24.6	26.9	50.3	0	6731+	1194	1270	3378
Ni-Cu Steel (1.96Ni-1.01Cu)	–	D	27.2	33.8	112.8	2977+	6299+	1422	1397	3505
Ni-Cu Steel (2.47Ni-1.08Cu)	40.9	37.3	9.5	26.7	D	0	3683+	559	2286	3683+
Ni-Cu Steel (1.95Ni-1.04Cu)	–	57.8	21.0	19.4	D	0	5334+	1219	1549	5334+
Cr-Si-Cu Steel (1.02Cr-0.42Cu)	–	D	24.6	22.6	D	0	4775+	1626	1473	3505+
2Cr Steel w/ Mo	–	D	20.0	21.6	D	0	3683+	2388	1778	3683+
2.01Cr-0.57Mo Steel	–	98.9	21.1	19.4	39.6	0	4445+	1092	1422	3531+
5.05Cr Steel	24.9	33.7	8.7	25.7	45.0	3404	3912	711	3226+	3480+
5.02Cr Steel	–	D	9.0	26.6	29.8	0	6223+	1397	3454	2057
4.67Cr-0.51Mo Steel	–	D	9.8	26.0	24.0	0	4775+	1168	2489	1524
5.76Cr-0.43Mo Steel	–	D	6.9	23.1	29.5	0	5156+	1397	2743	1397

2 Specimens

KEY: Average Penetration

+ One Specimen Destroyed By Corrosion

D Both Specimens Destroyed By Corrosion

Maximum Penetration

+ One or More Specimens Contained Holes

D Both Specimens Destroyed By Corrosion

Table V-A Corrosion Penetration for Galvanic Couples Exposed to Various Waters in the Panama Canal Zone for Periods of up to 16 Years

(Southwell & Alexander, 1976)

Strip Metal

Average Penetration (µm)

Strip	Plate	Sea Water				Mean Tide				Fresh Water			
(2x9x1/4 in)	(9x9x1/4 in)	1	8	16	16*	1	8	16	16*	1	8	16	16*
1 Carbon Steel (0.24% C)	Carbon Steel (0.24% C)	190.5	670.6	1270.0	1221.7	198.1	398.8	1198.9	1148.1	215.9	510.5	657.9	660.4
2 316 Stainless Steel	Carbon Steel (0.24% C)	0.0	0.0	0.0	50.8	2.5	0.0	2.5	5.1	0.0	0.0	0.0	0.0
3 302 Stainless Steel	Carbon Steel (0.24% C)	0.0	0.0	5.1	236.2	2.5	0.0	2.5	40.6	0.0	0.0	0.0	0.0
4 Nickel (99%)	Carbon Steel (0.24% C)	0.0	0.0	2.5	490.2	0.0	0.0	2.5	111.8	0.0	0.0	0.0	0.0
5 70Ni-30Cu (cold Rolled)	Carbon Steel (0.24% C)	0.0	0.0	2.5	221.0	0.0	0.0	2.5	68.6	0.0	0.0	0.0	15.2
6 70Cu-30Ni	Carbon Steel (0.24% C)	0.0	0.0	2.5	58.4	0.0	0.0	2.5	20.3	0.0	0.0	2.5	33.0
7 Aluminum Bronze (5% Al)	Carbon Steel (0.24% C)	2.5	2.5	10.2	33.0	2.5	2.5	10.2	15.2	0.0	2.5	5.1	10.2
8 Low Brass (80Cu-20Zn)	Carbon Steel (0.24% C)	2.5	2.5	10.2	94.0	2.5	2.5	12.7	17.8	2.5	5.1	5.1	43.2
9 Naval Brass (39Zn-1Sn)	Carbon Steel (0.24% C)	2.5	2.5	5.1	315.0	5.1	2.5	7.6	152.4	2.5	2.5	5.1	48.3
10 Copper (99%)	Carbon Steel (0.24% C)	50.8	76.2	177.8	152.4	50.8	76.2	152.4	33.0	0.0	25.4	76.2	25.4
11 Lead (99%)	Carbon Steel (0.24% C)	17.8	15.2	27.9	355.6	15.2	38.1	99.1	210.8	5.1	35.6	61.0	83.8
12 Nickel Steel (5% Ni)	Carbon Steel (0.24% C)	94.0	233.7	688.3	1310.6	99.1	198.1	673.1	1371.6	114.3	327.7	414.0	756.9
13 Aluminum (AA1100: 99% Al)	Carbon Steel (0.24% C)	678.2	3329.9	M	25.4a	607.1	2123.4	M	12.7a	10.2	271.8	734.1	127.0
14 Wrought Iron (Ashton Process)	Carbon Steel (0.24% C)	299.7	1021.1	1841.5	563.9a	233.7	442.0	1125.2	416.6a	200.7	584.2	762.0	370.8a
15 Zinc (99.5%)	Carbon Steel (0.24% C)	800.1	3761.7	4968.2	378.5	947.4	2603.5	4932.7	335.3	68.6	614.7	1094.7	200.7
16 Phosphor Bronze (4Sn-0.25P)	Carbon Steel (0.24% C)	2.5	5.1	10.2	139.7	5.1	2.5	10.2	88.9	2.5	2.5	2.5	17.8
Carbon Steel (0.24% C)	Carbon Steel (0.24% C)	190.5	670.6	1270.0	1221.7	198.1	398.8	1198.9	1148.1	215.9	510.5	657.9	660.4
Carbon Steel (0.24% C)	Aluminum (AA1100: 99% Al)	12.7	12.7	15.2	1221.7	17.8	20.3	58.4	1148.1	221.0	337.8	403.9	660.4
Carbon Steel (0.24% C)	Aluminum (6061 T)	12.7	20.3	40.6	1221.7	17.8	30.5	63.5	1148.1	231.1	403.9	431.8	660.4
Carbon Steel (0.24% C)	Galvanized Steel	7.6	259.1	ND	647.7a	33.0	185.4	ND	589.3a	88.9	276.9	ND	556.3a
Carbon Steel (0.24% C)	Wrought Iron (Ashton Process)	134.6	434.3	ND	647.7a	203.2	436.9	ND	589.3a	203.2	467.4	ND	556.3a
Carbon Steel (0.24% C)	0.3% Cu Steel	203.2	73.7	1033.8	1221.7	248.9	421.6	866.1	1148.1	185.4	525.8	602.0	660.4
Carbon Steel (0.24% C)	2% Ni Steel	271.8	843.3	1602.7	1221.7	274.3	589.3	1267.5	1148.1	248.9	729.0	861.1	660.4
Carbon Steel (0.24% C)	Cast Bronze (ounce metal)	937.3	3896.4	M	647.7a	619.8	2626.4	M	589.3a	381.0	1491.0	1963.4	660.4
Carbon Steel (0.24% C)	Cast Bronze (valve metal)	914.4	4699.0	M	647.7a	505.5	2059.9	M	589.3a	378.5	1440.2	1752.6	660.4
Carbon Steel (0.24% C)	Cast Tin Bronze (9% Sn)	894.1	4165.6	M	647.7a	533.4	2209.8	M	589.3a	383.5	1320.8	1549.4	660.4
Carbon Steel (0.24% C)	Cast Nickel-Tin Bronze (6Ni-3Sn)	924.6	4592.3	M	647.7a	533.4	2471.4	M	589.3a	370.8	1247.1	1518.9	660.4
Carbon Steel (0.24% C)	Muntz Brass (40Zn-0.25As)	944.9	4109.7	M	647.7a	922.0	3271.5	M	589.3a	370.8	1559.6	1966.0	660.4
Carbon Steel (0.24% C)	Manganese Brass (41Zn+Sn+Fe)	883.9	4031.0	M	647.7a	830.6	2697.5	M	589.3a	348.0	1165.9	1686.6	660.4
Carbon Steel (0.24% C)	Naval Brass (39Zn-1Sn)	894.1	4483.1	M	647.7a	878.8	2969.3	M	589.3a	353.1	1211.6	1633.2	660.4
Carbon Steel (0.24% C)	Cartridge Brass (30% Zn)	916.9	4122.4	M	647.7a	975.4	2966.7	M	589.3a	378.5	1447.8	2021.8	660.4
Carbon Steel (0.24% C)	Low Brass (20% Zn)	985.5	3860.8	M	647.7a	716.3	2656.8	M	589.3a	393.7	1290.3	1610.4	660.4
Carbon Steel (0.24% C)	Commercial Bronze (10%Zn)	962.7	2936.2	M	647.7a	541.0	2385.1	M	589.3a	358.1	1292.9	1719.6	660.4
Carbon Steel (0.24% C)	Aluminum Bronze (5% Al)	886.5	4315.5	M	647.7a	609.6	2608.6	M	589.3a	370.8	1445.3	1798.3	660.4
Carbon Steel (0.24% C)	Phosphor Bronze (4Sn-0.25P)	891.5	4371.3	M	647.7a	640.1	2443.5	M	589.3a	363.2	1158.2	1663.7	660.4
Carbon Steel (0.24% C)	Silicon Bronze (2.5% Si)	904.2	4511.0	M	647.7a	546.1	2446.0	M	589.3a	391.2	1455.4	1851.7	660.4
Carbon Steel (0.24% C)	Copper (99%)	927.1	4864.1	M	647.7a	447.0	2138.7	M	589.3a	360.7	1430.0	1973.6	660.4
Carbon Steel (0.24% C)	70Cu-30Ni Steel	952.5	4287.5	M	647.7a	589.3	2456.2	M	589.3a	353.1	1270.0	1742.4	660.4
Carbon Steel (0.24% C)	70Ni-30Cu Cold Rolled	868.7	4005.6	M	647.7a	591.8	2341.9	4864.1	1148.1	322.6	1016.0	1546.9	711.2
Carbon Steel (0.24% C)	Nickel (99%)	1003.3	4447.5	M	647.7a	515.6	2194.6	4693.9	1148.1	299.7	1219.2	1811.0	660.4
Carbon Steel (0.24% C)	410 Stainless Steel (13% Cr)	833.1	4061.5	M	647.7a	563.9	1872.0	4531.4	1148.1	299.7	1148.1	1432.6	660.4
Carbon Steel (0.24% C)	302 Stainless Steel (18Cr-8Ni)	881.4	3718.6	M	647.7a	576.6	2258.1	4460.2	1148.1	299.7	1043.9	1338.6	660.4
Carbon Steel (0.24% C)	316 Stainless Steel (18Cr-13Ni+Mo)	881.4	3571.2	M	647.7a	574.0	2080.3	4455.2	1148.1	266.7	845.8	1127.8	660.4
Zinc (99.5%)	Carbon Steel (0.24% C)	800.1	3761.7	4968.2	378.5	947.4	2603.5	4932.7	335.3	68.6	614.7	1094.7	200.7
Zinc (99.5%)	0.3% Cu Steel	810.3	4000.5	ND	231.1a	947.4	2321.6	ND	208.3a	73.7	574.0	ND	142.2a
Zinc (99.5%)	2% Ni Steel	769.6	3512.8	M	231.1a	1333.5	2128.5	4864.1	335.3	73.7	629.9	929.6	200.7
Zinc (99.5%)	Cast Steel (0.2% C)	779.8	3693.2	5016.5	231.1a	1323.3	2811.8	4927.6	335.3	68.6	581.7	1158.2	200.7
Zinc (99.5%)	Austenitic Cast Iron (18% Ni)	947.4	3030.2	M	231.1a	490.2	2585.7	3208.0	335.3	127.0	353.1	602.0	200.7
Phosphor Bronze (4Sn-0.25P)	Carbon Steel (0.24% C)	2.5	5.1	10.2	139.7	5.1	2.5	10.2	88.9	2.5	2.5	2.5	17.8
Phosphor Bronze (4Sn-0.25P)	0.3% Cu Steel	5.1	5.1	ND	101.6a	2.5	5.1	ND	53.3a	2.5	5.1	ND	12.7a
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cu-Ni)	2.5	2.5	12.7	139.7	2.5	2.5	10.2	88.9	0.0	2.5	2.5	17.8
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cu-Cr-Si)	5.1	7.6	12.7	139.7	2.5	2.5	7.6	88.9	2.5	2.5	5.1	17.8
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cu-Ni-Mn-Mo)	2.5	5.1	15.2	139.7	2.5	5.1	10.2	88.9	2.5	2.5	2.5	17.8
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cr-Ni-Mn)	2.5	5.1	12.7	139.7	2.5	2.5	7.6	88.9	2.5	2.5	2.5	17.8
Phosphor Bronze (4Sn-0.25P)	Nickel Steel (2% Ni)	2.5	5.1	12.7	139.7	2.5	2.5	10.2	88.9	0.0	2.5	5.1	17.8
Phosphor Bronze (4Sn-0.25P)	Chromium Steel (2% Cr)	5.1	7.6	15.2	139.7	5.1	7.6	10.2	88.9	2.5	2.5	5.1	17.8
Phosphor Bronze (4Sn-0.25P)	Cast Steel (0.27% C)	2.5	5.1	5.1	139.7	2.5	2.5	7.6	88.9	2.5	5.1	2.5	17.8
Phosphor Bronze (4Sn-0.25P)	Gray Cast Iron (3.2% C)	2.5	5.1	7.6	139.7	0.0	5.1	7.6	88.9	0.0	5.1	2.5	17.8
Phosphor Bronze (4Sn-0.25P)	Austenitic Cast Iron (18% Ni)	2.5	2.5	7.6	139.7	2.5	2.5	5.1	88.9	15.2	5.1	0.0	17.8
Phosphor Bronze (4Sn-0.25P)	70Cu-30Ni	40.6	83.8	116.8	139.7	17.8	33.0	40.6	88.9	2.5	10.2	12.7	17.8
Phosphor Bronze (4Sn-0.25P)	70Ni-30Cu (cold Rolled)	68.6	1684.0	1823.7	139.7	43.2	378.5	482.6	88.9	40.6	223.5	358.1	17.8

Table V-A Corrosion Penetration for Galvanic Couples Exposed to Various Waters in the Panama Canal Zone for Periods of up to 16 Years

(Southwell & Alexander, 1976) – (Continued)

Plate Metal

Average Penetration (μm)

Strip	Plate	Sea Water				Mean Tide				Fresh Water			
(2x9x1/4 in)	(9x9x1/4 in)	1	8	16	16*	1	8	16	16*	1	8	16	16*
Carbon Steel (0.24% C)	Carbon Steel (0.24% C)	172.7	609.6	1117.6	1221.7	248.9	454.7	1071.9	1148.1	213.4	520.7	670.6	660.4
316 Stainless Steel	Carbon Steel (0.24% C)	200.7	718.8	1257.3	1221.7	309.9	627.4	1110.0	1148.1	203.2	637.5	823.0	660.4
302 Stainless Steel	Carbon Steel (0.24% C)	190.5	693.4	1285.2	1221.7	297.2	584.2	1198.9	1148.1	205.7	635.0	805.2	660.4
Nickel (99%)	Carbon Steel (0.24% C)	195.6	759.5	1290.3	1221.7	304.8	543.6	1074.4	1148.1	195.6	602.0	739.1	660.4
70Ni-30Cu (cold Rolled)	Carbon Steel (0.24% C)	203.2	706.1	1300.5	1221.7	266.7	584.2	1087.1	1148.1	200.7	591.8	675.6	660.4
70Cu-30Ni	Carbon Steel (0.24% C)	193.0	756.9	1280.2	1221.7	312.4	533.4	1346.2	1148.1	200.7	579.1	756.9	660.4
Aluminum Bronze (5% Al)	Carbon Steel (0.24% C)	188.0	726.4	1463.0	1221.7	269.2	602.0	1170.9	1148.1	215.9	586.7	711.2	660.4
Low Brass (80Cu-20Zn)	Carbon Steel (0.24% C)	188.0	741.7	1381.8	1221.7	304.8	508.0	1125.2	1148.1	218.4	632.5	746.8	660.4
Naval Brass (39Zn-1Sn)	Carbon Steel (0.24% C)	182.9	754.4	1381.8	1221.7	340.4	579.1	1145.5	1148.1	218.4	604.5	711.2	660.4
Copper (99%)	Carbon Steel (0.24% C)	193.0	749.3	1394.5	1221.7	302.3	584.2	1084.6	1148.1	198.1	607.1	777.2	660.4
Lead (99%)	Carbon Steel (0.24% C)	182.9	734.1	1231.9	1221.7	292.1	602.0	1112.5	1148.1	203.2	591.8	779.8	660.4
Nickel Steel (5% Ni)	Carbon Steel (0.24% C)	175.3	706.1	1201.4	1221.7	289.6	528.3	1183.6	1148.1	223.5	538.5	734.1	660.4
Aluminum (AA1100: 99% Al)	Carbon Steel (0.24% C)	17.8	15.2	429.3	1221.7	20.3	33.0	61.0	1148.1	188.0	472.4	551.2	660.4
Wrought Iron (Ashton Process)	Carbon Steel (0.24% C)	157.5	556.3	939.8	1221.7	254.0	500.4	1092.2	1148.1	0.0	0.0	0.0	0.0
Zinc (99.5%)	Carbon Steel (0.24% C)	17.8	22.9	393.7	1221.7	27.9	22.9	50.8	1148.1	177.8	452.1	505.5	660.4
Phosphor Bronze (4Sn-0.25P)	Carbon Steel (0.24% C)	190.5	751.8	1399.5	1221.7	287.0	566.4	1178.6	1148.1	221.0	622.3	734.1	660.4
Carbon Steel (0.24% C)	Carbon Steel (0.24% C)	172.7	609.6	1117.6	1221.7	248.9	454.7	1071.9	1148.1	213.4	520.7	670.6	660.4
Carbon Steel (0.24% C)	Aluminum (AA1100: 99% Al)	38.1	129.5	180.3	25.4	25.4	63.5	167.6	12.7	10.2	132.1	292.1	127.0
Carbon Steel (0.24% C)	Aluminum (6061 T)	45.7	180.3	198.1	22.9	25.4	58.4	139.7	7.6	5.1	109.2	238.8	38.1
Carbon Steel (0.24% C)	Galvanized Steel												
Carbon Steel (0.24% C)	Wrought Iron (Ashton Process)	188.0	627.4	ND	563.9a	177.8	459.7	ND	416.6a	213.4	502.9	ND	370.8a
Carbon Steel (0.24% C)	0.3% Cu Steel	165.1	655.3	1064.3	1252.2	266.7	599.4	1150.6	1173.5	190.5	528.3	637.5	729.0
Carbon Steel (0.24% C)	2% Ni Steel	139.7	629.9	1224.3	1320.8	177.8	518.2	1193.8	1404.6	193.0	421.6	485.1	558.8
Carbon Steel (0.24% C)	Cast Bronze (ounce metal)	0.0	2.5	78.7	104.1	2.5	5.1	7.6	30.5	0.0	2.5	7.6	10.2
Carbon Steel (0.24% C)	Cast Bronze (valve metal)	2.5	2.5	48.3	55.9	2.5	2.5	2.5	35.6	0.0	2.5	5.1	10.2
Carbon Steel (0.24% C)	Cast Tin Bronze (9% Sn)	2.5	2.5	63.5	137.2	2.5	5.1	5.1	35.6	0.0	2.5	5.1	7.6
Carbon Steel (0.24% C)	Cast Nickel-Tin Bronze (6Ni-3Sn)	0.0	2.5	40.6	152.4	0.0	2.5	5.1	30.5	0.0	5.1	7.6	12.7
Carbon Steel (0.24% C)	Muntz Brass (40Zn-0.25As)	2.5	2.5	132.1	208.3	5.1	7.6	22.9	144.8	5.1	25.4	38.1	88.9
Carbon Steel (0.24% C)	Manganese Brass (41Zn+Sn+Fe)	2.5	2.5	200.7	436.9	5.1	7.6	38.1	190.5	2.5	7.6	10.2	22.9
Carbon Steel (0.24% C)	Naval Brass (39Zn-1Sn)	2.5	2.5	188.0	315.0	7.6	7.6	17.8	152.4	2.5	12.7	20.3	48.3
Carbon Steel (0.24% C)	Cartridge Brass (30% Zn)	0.0	0.0	78.7	91.4	2.5	2.5	7.6	22.9	2.5	7.6	10.2	45.7
Carbon Steel (0.24% C)	Low Brass (20% Zn)	2.5	2.5	61.0	94.0	2.5	2.5	7.6	17.8	2.5	5.1	7.6	43.2
Carbon Steel (0.24% C)	Commercial Bronze (10%Zn)	2.5	2.5	78.7	124.5	0.0	2.5	5.1	25.4	0.0	5.1	5.1	33.0
Carbon Steel (0.24% C)	Aluminum Bronze (5% Al)	0.0	0.0	101.6	33.0	0.0	2.5	5.1	15.2	0.0	2.5	5.1	10.2
Carbon Steel (0.24% C)	Phosphor Bronze (4Sn-0.25P)	2.5	2.5	48.3	139.7	2.5	5.1	5.1	88.9	2.5	5.1	7.6	17.8
Carbon Steel (0.24% C)	Silicon Bronze (2.5% Si)	2.5	5.1	68.6	149.9	2.5	5.1	7.6	50.8	2.5	5.1	7.6	30.5
Carbon Steel (0.24% C)	Copper (99%)	2.5	2.5	73.7	152.4	2.5	2.5	12.7	33.0	2.5	2.5	5.1	25.4
Carbon Steel (0.24% C)	70Cu-30Ni Steel	0.0	2.5	40.6	58.4	0.0	2.5	2.5	20.3	0.0	5.1	7.6	33.0
Carbon Steel (0.24% C)	70Ni-30Cu Cold Rolled	0.0	2.5	61.0	221.0	0.0	2.5	2.5	68.6	0.0	0.0	0.0	15.2
Carbon Steel (0.24% C)	Nickel (99%)	0.0	0.0	121.9	482.6	0.0	2.5	7.6	111.8	0.0	0.0	5.1	0.0
Carbon Steel (0.24% C)	410 Stainless Steel (13% Cr)	5.1	2.5	233.7	607.1	2.5	2.5	12.7	162.6	0.0	0.0	0.0	22.9
Carbon Steel (0.24% C)	302 Stainless Steel (18Cr-8Ni)	0.0	0.0	20.3	241.3	0.0	0.0	0.0	40.6	0.0	0.0	0.0	0.0
Carbon Steel (0.24% C)	316 Stainless Steel (18Cr-13Ni+Mo)	0.0	0.0	0.0	50.8	0.0	0.0	0.0	2.5	0.0	0.0	0.0	0.0
Zinc (99.5%)	Carbon Steel (0.24% C)	17.8	22.9	393.7	1221.7	27.9	22.9	50.8	1148.1	177.8	452.1	505.5	660.4
Zinc (99.5%)	0.3% Cu Steel	10.2	96.5	ND	1252.2a	25.4	17.8	ND	1173.5a	215.9	475.0	ND	729.0a
Zinc (99.5%)	2% Ni Steel	10.2	7.6	378.5	1320.8	43.2	22.9	66.0	1404.6	170.2	381.0	442.0	579.1
Zinc (99.5%)	Cast Steel (0.2% C)	10.2	10.2	289.6	1099.8	38.1	25.4	129.5	1247.1	188.0	421.6	556.3	668.0
Zinc (99.5%)	Austenitic Cast Iron (18% Ni)	0.0	0.0	2.5	579.1	0.0	0.0	0.0	185.4	17.8	99.1	195.6	203.2
Phosphor Bronze (4Sn-0.25P)	Carbon Steel (0.24% C)	190.5	751.8	1399.5	1221.7	287.0	566.4	1178.6	1148.1	221.0	622.3	734.1	660.4
Phosphor Bronze (4Sn-0.25P)	0.3% Cu Steel	198.1	828.0	ND	1252.2a	360.7	655.3	ND	1173.5a	203.2	584.2	ND	729.0a
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cu-Ni)	188.0	731.5	1353.8	1229.4	332.7	876.3	1422.4	1247.1	170.2	457.2	584.2	558.8
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cu-Cr-Si)	205.7	1115.1	1993.9	2049.8	274.3	657.9	1518.9	1252.2	157.5	459.7	612.1	619.8
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cu-Ni-Mn-Mo)	188.0	675.6	1226.8	1145.5	284.5	538.5	1430.0	1295.4	193.0	576.6	807.7	792.5
Phosphor Bronze (4Sn-0.25P)	Low Alloy Steel (Cr-Ni-Mn)	203.2	1143.0	2529.8	2136.1	251.5	617.2	1656.1	1163.3	152.4	416.6	647.7	680.7
Phosphor Bronze (4Sn-0.25P)	Nickel Steel (2% Ni)	200.7	810.3	1544.3	1320.8	218.4	596.9	1419.9	1397.0	208.3	551.2	624.8	535.9
Phosphor Bronze (4Sn-0.25P)	Chromium Steel (2% Cr)	55.9	1102.4	2197.1	1788.2	276.9	754.4	1376.7	1221.7	76.2	284.5	411.5	373.4
Phosphor Bronze (4Sn-0.25P)	Cast Steel (0.27% C)	210.8	756.9	1310.6	1099.8	254.0	574.0	1231.9	1247.1	208.3	462.3	764.5	668.0
Phosphor Bronze (4Sn-0.25P)	Gray Cast Iron (3.2% C)	254.0	1508.8	2659.4	2400.3	492.8	1315.7	1935.5	2212.3	134.6	596.9	845.8	835.7
Phosphor Bronze (4Sn-0.25P)	Austenitic Cast Iron (18% Ni)	106.7	383.5	693.4	579.1	48.3	127.0	259.1	185.4	33.0	177.8	337.8	203.2
Phosphor Bronze (4Sn-0.25P)	70Cu-30Ni	15.2	50.8	96.5	58.4	2.5	15.2	25.4	20.3	5.1	25.4	33.0	33.0
Phosphor Bronze (4Sn-0.25P)	70Ni-30Cu (cold Rolled)	7.6	50.8	177.8	221.0	2.5	25.4	55.9	68.6	0.0	10.2	20.3	15.2

Table V-B Plates Which, In the Following Environments, Experienced Increased Corrosion Rates When Coupled With Carbon Steel Plates

(Southwell, et al., 1976)

Sea Water	Mean Tide	Fresh Water
99% Al	99% Al	99% Al
6061T Al Alloy	6061T Al Alloy	6061T Al Alloy
5% Al Bronze	Wrought Iron	Wrought Iron (slightly)

Table V-C Corrosion Rate of AISI 4340 Steel When Coupled With Various Metals (µm/yr) (Pelensky, et al., 1976; 1978)

	Atmosphere (24 month)	Soil (25 month)	Sea water (18 month)
4340 steel	0.048	0.580	0.338
Az31 Mg	0.011	0.029	Missing
316 Stainless Steel	0.049	0.490	0.596
6061 Al Alloy	0.026	0.056	0.005
7075 Al Alloy	0.025	0.109	0.004
360 Brass	0.037	0.561	0.504
Monel 400	0.049	0.603	0.535
Ti-6Al-4V	0.047	0.497	0.589

Table V-D Corrosion Rate of Various Metals When Coupled With AISI 4340 Steel (μm/yr) (Pelensky, et al., 1976; 1978)

	Atmosphere (24 month)		Soil (25 month)		Sea water (18 month)	
Couple :	Itself	4340	Itself	4340	Itself	4340
Az31 Mg	0.028	0.076	0.029	0.580	Missing	Missing
6061 Al Alloy	0.001	0.007	0.001	0.003	0.006	0.081
7075 Al Alloy	0.001	0.012	0.005	0.002	0.049	0.113
360 Brass	0.001	0.002	0.004	0.0	0.115	0.001
Monel 400	0.0	0.0	0.004	0.0	0.027	0.001
316 Stainless	0.0	0.0	0.0	0.0	0.551	0.0