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REPORT

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MASTER

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On.....A STUDY OF THE MECHANISM OF STRESS CORROSION
CRACKING IN THE IRON-NICKEL-CHROMIUM ALLOY
SYSTEM

For the period.....17 March 1964 - 16 June 1964

Submitted by.....F. H. Beck, M. G. Fontana, and J. P. Hirth
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I. INTRODUCTION

The aim of this program is to evaluate systematically the susceptibility of iron-chromium-nickel alloys to stress corrosion cracking in chloride- and hydroxide-containing environments. The general experimental program will map electrochemical, physical, and corrosion properties of specially prepared high purity alloys in the α Fe, γ Ni and α - γ two-phase regions. Standard commercial stainless steels and high nickel alloys are being studied also. The over-all plan and range of compositions to be studied are described in previous progress reports.

II. LITERATURE SURVEY

Attached (Appendix A) is a preliminary literature survey on the subject of stress corrosion cracking. This literature survey emphasizes some of the important contributions in the area of defect structure and electrochemistry. A detailed analyses of possible mechanisms will be presented when sufficient experimental work from this program is available to verify various important conclusions.

III. FABRICATION AND MATERIALS

A. PREPARATION OF MASTER ALLOYS

Ingots of specific compositions will be prepared from master alloys now being melted. A summary of results from chemical analyses of the master alloys is given in Table I.

Analyses of the starting materials is shown in Table II.

TABLE I. COMPOSITION OF MASTER ALLOYS

Alloy Melt No.	Nominal						Analyzed ⁽¹⁾ ⁽²⁾						
	Atom Percent			Weight Percent			Weight Percent						
	Fe	Cr	Ni	Fe	Cr	Ni	Fe	Cr	Ni	C	O	H	N
1	95	5	-	95.3	4.7	-	95.09	4.82	-	.0120	.0015	.00004	.0010
2	85	15	-	85.9	14.1	-	85.53	14.22	-	.0110	.0028	.00004	.0020
3	70	30	-	71.5	28.5	-	71.25	28.49	-	.0050	.0100	<.00003	.0030
4	-	5	95	-	4.5	95.5	0.72	4.88	94.50	.0060	.0027	.00003	<.0010
5	-	20	80	-	18.1	81.9	1.12	7.57	81.00	.0060	.0019	.00005	.0010
6	-	30	70	-	27.5	72.5	3.91	26.70	68.90	.0020	.0051	.00004	.0030
7	45	5	50	44.0	4.6	51.4	42.70	4.83	52.50	.0020	.0048	.00005	.0010
8	8	15	77	7.8	13.6	78.6	8.70	13.57	77.50	.0040	.0029	.00003	.0020
8-1	8	15	77	7.8	13.6	78.6	9.0	13.40	Bal	.0030	.0042	.00003	.0010
9	75	20	5	75.9	18.8	5.3	75.45	19.55	5.20	.0030	.0042	.00006	.0030
10	65	20	15	65.4	18.7	15.9	65.40	19.38	15.13	.0060	.0035	.00004	.0020
10-1	65	20	15	65.4	18.7	15.9	Bal	19.38	15.83	.0100	.0047	.00003	.0030
11	25	30	45	24.9	27.9	47.2	25.16	27.70	47.30	.0020	.0084	.00003	.0030

(1) Analyses by The Duriron Company, Dayton, Ohio for Fe, Cr, Ni, X-ray Quantometer and wet chemical. All numbers shown indicate direct analyses. Where "bal" shown, the composition is obtained by difference.

(2) Interstitial analyses by Battelle Memorial Institute: Oxygen and hydrogen by vacuum fusion. Nitrogen by Kjeldahl technique.

TABLE II. ANALYSES OF STARTING MATERIALS (%)

Element	Glidden 104 Iron	ELCHROME HP CHROMIUM		Ni-270 Nickel
		1st Batch	2nd Batch	
C	0.005	0.008	-	0.020
N	0.004	0.010	0.0070	0.00006*
O	0.050	0.013	0.0060	--
H	--	0.0001	-	0.00009*
Mn	0.002	--	-	< 0.003
Fe	--	--	-	< 0.005
Cu	--	--	-	< 0.003
Cr	0.004	--	-	< 0.003
S	0.005	0.006	-	< 0.003
Si	0.005	0.007	-	< 0.003
Mg	--	--	-	< 0.003
Ti	--	--	-	< 0.003
Co	0.002	--	-	< 0.003
Ni	0.002	--	-	--
P	0.004	--	-	--
Element	< 99.9	99.95	99.9+	99.97

*BMI analysis, vacuum fusion technique

Comparing the information in Tables I and II, it is apparent that considerable reductions have been made in carbon, oxygen, and nitrogen levels. This has been achieved by vacuum induction melting and making appropriate additions of carbon (as iron carbide) or additions of oxygen (as iron oxide). It will be noted that in several cases the oxygen or carbon is relatively higher than the average; but it is expected that these will be reduced when the one pound ingots are melted.

As of June 17, 1964, approximately 30 of the one-pound ingots have been melted and 18 of these have been fabricated to 1/8 inch diameter.

B. PREPARATION OF FOURTH COMPONENT ALLOYS

At five locations in the ternary diagram (see Fig. 1) it is planned to add fourth component alloying additions. This is based on numerous observations that certain fourth components have important effects on stress corrosion cracking. The significance of the various trends is not completely clear but tentative suggestions are related to the effect of the fourth component on:

1. Hydrogen pickup by the metal.
2. Stacking fault energy.
3. Locking of line defects.
4. Local anodic regions due to:
 - a) second phase precipitates,
 - b) thermodynamic enrichment of component at grain boundary region, and
 - c) enrichment at stacking faults or at dislocations.
5. Reduction in energy required to form new surfaces.

It is apparent that there are a number of fruitful avenues for analysis. In order to provide a useful fund of information for analysis of the effects of fourth component alloy additions, it was decided to melt alloys of all the fourth components which exhibit appreciably solid solubility individually in iron, chromium, and nickel. Table III lists the alloy elements which will be added and appropriate solid solubilities in α iron, γ iron, nickel and chromium.

Each of the alloys in Table III will be added to composition 54 (65 Fe-20 Cr-15 Ni) at the 1.5 atomic percent level. (see Fig. 1) In addition carbon, oxygen, nitrogen, phosphorus, and sulfur will be added to composition 54 at the 0.1 atomic percent level. The primary investigation of fourth component additions will be made at composition 54. A limited number of fourth component additions will be made at compositions 21 (45 Fe-50 Ni-5 Cr), 50 (8 Fe-77 Ni - 15 Cr), 52 (75 Fe-5 Ni-20 Cr), and 78 (25 Fe-45 Ni-30 Cr) to serve as comparisons. The fourth component elements in these latter four ternary bases will be added at the 1.5 atomic percent level. Thus at the different ternary bases the effect of a given fourth component can be compared directly. When the elements carbon, sulfur, phosphorous, nitrogen, or oxygen are added, it will be at the 0.1 atomic percent level at the 21, 50, 52, and 78 compositions. In the first round of one-pound ingots only the following fourth component elements will be added at the indicated ternary bases: 21: Be, Mo, S, N; 50: Al, Be, Mo, Ti, N, P, S; 52: Be, Mo, S, N; 78: Be, Mo, N, S. These additions have been selected to evaluate melting problems and provide a preliminary comparison at alloy bases different from number 54.

Figure 1

ALLOYS TO BE MELTED
(400° C ISOTHERM SHOWN)

TABLE III. FOURTH COMPONENT ALLOYING ELEMENTS AND IMPORTANT ALLOYING DATA (1)

Element	Solubility (atomic percent)(2)				Distance of Closest Approach (Å)	Crystal Structure (in temperature range of interest)
	α-Fe(3)	γ-Fe(4)	Ni	Cr		
Al	(18.5)	2	10	(27)	2.862	fcc
As	(7)	(2)	(2)	(1.4)	2.51	Rhombohedral
Au	0	8	1	(7)	2.884	fcc
Be	(9)	(2)	2	2	2.225	hcp
Co	(72.0)	100	90	9	2.506	hcp < 400°C; fcc > 400°C
Cu	(3)	7.5	100	0	2.556	fcc
Ir	(9)	100	-	(2)	2.714	fcc
Mn	3	100	(10)	(25)	2.24	cubic
Mo	4	2	(12)	100	2.725	bcc
Os	(5)	(large)	(5)	(9.5)	2.675	hcp
Pd	(1)	100	100	(5)	2.750	fcc
Pt	(12)	100	(10)	(2)	2.775	fcc
Re	10	16.7	-	?	2.740	hcp
Rh	50	(30)	-	(8)	2.689	fcc
Ru	9	(large)	12	(15)	2.649	hcp
Sb	(1)	(1)	4	(6)	2.903	Rhombohedral
Si	10	4	10	(5)	2.351	cubic
Ti	(3)	(2)	8	(1)	2.89	hcp
V	(22)	1.6	15	(100)	2.632	bcc
W	(2)	(1)	12	(3)	2.739	bcc
Zn	6	42.2	(10)	(25.5)	2.664	hcp
Fe	-	-	-	-	2.481	bcc
Cr	-	-	-	-	2.498	bcc
Ni	-	-	-	-	2.491	fcc

(1) All information on solubility taken from Hansen, Constitution of Binary Alloys

(2) Numbers in parentheses are based on dotted lines or less certain data

(3) Taken at 400°C.

(4) Taken at maximum

For clarity, the alloy elements being considered as fourth component additions are shown in Fig. 2.

The distance of closest approach for the various fourth component additions is shown in Fig. 3 as compared with the $\pm 15\%$ range on the iron chromium and nickel. It is seen that these elements fall generally within the prescribed range for solid solubility.

IV. STRESS CORROSION CRACKING EXPERIMENTS

A. RECIRCULATING AUTOCLAVE SYSTEM

The components for the recirculating system have been fabricated and a final check out of the pump and the oxygen injection system will be conducted prior to delivery in the vendor's plant on June 18 and 19. Following this the components will be shipped to Ohio State University for installation.

B. STRESSING JIG PROTOTYPE

Thirty-five stressing jigs to be used in the small autoclaves and the $MgCl_2$ cell have been completed and preliminary experiments utilizing the jigs will begin shortly. If satisfactory results are obtained, about 250 jigs will be constructed for use in the recirculating autoclave.

The design of the stressing jig to be used is shown in Fig. 4. Also shown in Fig. 4 is the schematic procedure for stressing wire specimens.

There are a number of considerations in designing the stressing jig:

1. Specimen should crack in a relatively short time.
2. There should be a positive indication of the crack.
3. The applied force should be adjustable and controllable.
4. The jig should permit monitoring the time at which the specimen breaks.

The types of stressing arrangement considered and their peculiar problems are listed below:

<u>Stressing Technique</u>	<u>Problems</u>
1. Dead load	too large
2. Springs	
a) Torsion (specimen would be bow string on bow)	awkward and too large
b) Constant force (Ref. 1)	not possible to change force without changing spring

H																	He
Li	Be											B	C	N	O	F	Ne
Na	Mg											Al	Si	P	S	Cl	Ar
K	Ca	Sc	Ti	V	Cr	Mn	Fe	Co	Ni	Cu	Zn	Ga	Ge	As	Se	Br	Kr
Cs	Sr	Y	Zr	Nb	Mo	Tc	Ru	Rh	Pd	Ag	Cd	In	Sn	Sb	Te	I	Xe
Rb	Ba	La	Hf	Ta	W	Re	Os	Ir	Pt	Au	Hg	Tl	Pb	Bi	Po	At	Rn
Fr	Ra	Ac															

Fourth Component Alloying Elements Shown in Cross Hatch

Added at 1.5 Atom Percent Level

Added at 0.1 Atom Percent Level

Figure 2. Periodic Table showing elements to be added as fourth component additions

Figure 3. Comparison of atomic diameters of fourth component alloying elements with the $\pm 15\%$ limit of α -Fe, γ -Fe, Ni, and Cr

SCHEMATIC DIAGRAM OF SPECIMEN
STRESSING JIG

SCHEMATICS 1 and 2 IS THE
PROCEDURE FOR STRESSING
SPECIMENS IN SPRING JIG

Figure 4. Schematic of jig to be used for stressing specimens and the procedure for stressing

Stressing Technique

c) Compression

Problems

buckles if too great
L/D ratio

It was decided to choose the compression spring based primarily on size and ease of handling considerations.

Important considerations in the design of compression springs for the stressing jigs are listed below:

1. Minimum volume	-	maximum number of specimens
2. Minimum stress in springs	-	reduce relaxation and chemical attack
3. Maximum deflection to give desired stress	-	reduce errors in stress level
4. Restrict length of spring, unstressed	-	minimize buckling tendency
5. Minimize solid height of spring	-	improve circulation around specimen

Figure 5 shows deflection vs. load and deflection vs. stress diagrams for springs of 1.2-inch diameter with six to twelve active coils and spring wire diameters for 0.080 to 0.120 inch. From this type of curve and similar ones for other spring diameters it is possible to optimize the spring geometry.

Figure 6 (from Ref. 2) describes the region of spring geometry in which buckling will not occur. The springs to be used in this study are generally 2.0 inch long and 1.1 inch in diameter. The ratio l_0/D is on the order of two or less so that buckling would not be expected to occur. In order to evaluate the applicability of Fig. 6 to the geometry of the stressing jig in Fig. 4, a series of springs was fabricated with l_0/D ratios from 1.5 to 3.0 and $l_0 = 2.0$ inches. The spring with l_0/D at 3.0 was found to be unstable.

The maximum temperature at which the springs will be required to operate is 700°F. At this temperature a question is raised concerning the relaxation of the Inconel-600 spring material. Figure 7 (Ref. 3) shows the effect of temperature, stress, and time on relaxation. Approximately 2% relaxation at 700°F is considered acceptable.

Figure 5. Curves for deflection vs. load and stress in spring vs. deflection for spring diameter of 1.2 inch and for various numbers of active coils and spring wire diameter

Figure 6. Curves for evaluating the buckling tendencies of compression springs

Figure 7. Relaxation of Inconel-600 in the spring temper condition at various stresses, times, and temperatures

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APPENDIX A

PRELIMINARY SURVEY OF THE LITERATURE ON THE SUBJECT OF STRESS CORROSION CRACKING

For permission to reproduce the various figures and photographs the following are acknowledged.

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I. INTRODUCTION

Stress corrosion cracking is a mechanical-chemical process leading to the cracking of alloys at stresses considerably below the tensile strength. This phenomenon is very complex in that it depends intimately on the environment (phase, temperature, dissolved gas, ionic composition, velocity), metal (major alloying elements, impurities, crystal structure, fabrication history, distribution of phases), and stress level. It is clearly (insofar as responsible factors although sometimes not in metallographic appearance) different from other important forms of cracking as listed below:

1. Brittle fracture
2. Hydrogen embrittlement
3. Intergranular corrosion
4. Corrosion fatigue
5. Stress rupture

Stress corrosion cracks may proceed either transgranularly or intergranularly. Often within a single alloy system the mode of cracking will change according to changes in pH, heat treatment, applied potential, environment composition, or changes in alloy composition.

The strength level of an alloy in stress corrosion cracking varies widely depending on the environment and metal composition. Stresses to cause stress corrosion cracking in Type 304 stainless steel are compared with strength levels and cracking strengths produced by other methods of testing and in various methods of fabrication as shown in Figure 1.

To be classified as "stress corrosion cracking" the following ingredients must be present:

1. Tensile stress-- Only tensile stresses will lead to stress corrosion cracking
2. Environment--Stress corrosion cracking will not proceed unless an environment is present which will permit electrochemical reactions to proceed. Further, only certain environments have the requisite properties to sustain stress corrosion cracking, and then only in certain alloy systems.
3. Susceptible alloy--For a given alloy system in an environment which usually results in cracking of metals having the necessary stresses not all variations of the alloy constituents will result in cracking. Thus in the consideration of cracking phenomena, it is important to consider environment-alloy systems. In other words an environment does not universally cause stress corrosion cracking and a given alloy is not universally susceptible.

Stress Level, psi

Figure 1 Comparison of Fracture Stresses by Various Techniques. Compared with Stress Corrosion Cracking. Material: Type 304 Stainless

Stress corrosion cracking has received extensive attention at many laboratories because of a) the challenge of its complex behavior and b) the fact that many important commercial alloys which perform very well in some of the most severe chemical environments can be readily "done in" by an apparently innocuous environment. Brasses for example will crack in the presence of a small ammoniacal concentration. Certain stainless steels are subject to cracking in ordinary tap water.

Explanations for stress corrosion cracking until 1958 included the following:

1. Mechanical-electrochemical--A pit forms at the metal surface which leads to a stress concentration, this leads to local fracture, additional attack at the base of the fracture permits the progress of cracking, and the process is repeated.
2. Stress accelerated decomposition--The presence of severe local strains leads to the change of phase (in the case of Type 304 ferrite is produced from the austenitic structure). This phase would be anodic and corrode preferentially thus leading to new stress concentration and more strain. The process is then repeated.
3. Film rupture--It is hypothesized that plastic flow at the base of a crack will crack the protective oxide film. Where the film is broken will be an anodic region which will corrode. This leads to stress raisers and more strain to keep the film broken. This process continues.
4. Anodic shift--It has also been suggested that the primary factor leading to localized corrosion at the base of a crack is a higher corrosion potential produced by the higher stress level at the base of the crack.

While the above are generally reasonable they lack considerably in experimental verification and in accounting for the many diverse factors which are known to affect the course of stress corrosion cracking.

From 1958 to present a number of very important observations have been made in the area of electrochemistry, physical metallurgy, and stress corrosion testing which have considerably illuminated some of the dark corners of the stress corrosion cracking picture.

It is the purpose of this discussion to review many of these recent findings and to organize them into a perspective which may permit constructive analysis. Most of the examples cited herein are related to the iron-chromium-nickel alloy system and especially the austenitic stainless steels, but a number of examples have been drawn from other systems such as the brasses.

At the end of this paper is a brief bibliography and list of references which should enable the reader to contact most of the significant experimental work on stress corrosion cracking.

II. IMPORTANT EXPERIMENTAL CONTRIBUTIONS TO THE THEORY OF STRESS CORROSION CRACKING

A. Sources of Stress

The most important requirement of the stress is that it be tensile. This state of stress can be achieved in a number of ways. It is clear that such stresses may result from applied loads or from thermal stresses. Residual stresses due to welding, cold fabrication, local impacts on the metal, and machining may also contribute stresses sufficient to trigger stress corrosion cracks. Typical of cracks which result from stresses due to machining is the one shown in Figure 2. (1).*

Two additional sources of stress are outlined below:

1. Wedging Action of Corrosion Products--Pickering et al (2) have shown that corrosion products themselves can contribute stresses sufficient to cause the propagation of a crack without an applied load. This is demonstrated in Figure 3.
2. Stresses Due to Oxide Film--It has been demonstrated numerous times that the formation of an oxide film may lead to appreciable stresses on the surface of a metal. Chirigos (3) has suggested based on various experimental evidences that in the case of uranium base alloys the force exerted by mismatch between oxide and base metal is sufficient to produce cracking.

B. Stress Level

In general the time to cracking appears to decrease as the stress is increased. The general pattern of stress and time to cracking is demonstrated in Figure 4 (4). The existence of a minimum stress below which cracking does not occur has been the subject of extensive discussion. The apparent minimum stress below which cracking does not occur appears to be affected by alloy composition, the environment composition, and the physical phase of the medium. For example, in Ref. 1 it was demonstrated that when specimens are immersed in single phase aqueous solution cracking would not occur until applied stresses in excess of 20,000 psi were obtained whereas in a wetting and drying type of situation cracking would occur at stresses as low as 2000 psi. The obvious influence of stress on time to cracking probably indicates the necessity for a certain amount of plastic flow to occur before the cracks can initiate or propagate.

*Numbers in parentheses refer to references at the end of the paper

Figure 2 Stress corrosion cracks in Type 347 stainless steel due to stresses from machining. Vapor condensation conditions over a solution containing chlorides in the 0-50 ppm range. 350x

C. Alloy Composition and Microstructure

Summarized below are a number of important observations selected primarily from work in iron-chromium-nickel type alloys which give important clues to the role of compositional variations in affecting the susceptibility to stress corrosion cracking.

Figure 3 The wedging action of corrosion products. This crack in stainless steel has proceeded in its circular path under the influence of stresses produced only by corrosion products.

Figure 4 Composite curves illustrating the relative stress corrosion cracking resistance for commercial stainless steels in boiling 42 percent magnesium chloride

1. Effect of Nickel in Iron-Chromium-Nickel Alloys

The curve in Figure 5 developed by Copson (5) is generally typical of the effect of nickel in iron-chromium-nickel alloys.

A similar trend was demonstrated to exist by Copson for nickel additions to an iron base and is shown in the following table.

Figure 5 Stress corrosion cracking of iron-chromium-nickel wires in boiling 42% magnesium chloride

Table I

Stress Corrosion Cracking of Iron-Nickel U-Bends in Magnesium Chloride

Composition of Material Ni, %	Time to Crack, Days	Comment
Ni1	none ^a	
2.02	7	Few shallow cracks
4.96	-3	Profuse deep cracks
8.67	-3	Cracked in two
27.88	none ^b	
41.79	none ^b	
99.41	none ^b	

a-no crack in 11 days
b-no crack in 7 days

2. Pure Metals

The consensus of investigators in the field is that pure metals do not crack. The literature verifying this is well summarized by Barnartt (6).

3. Cracking of Alloys

Alloys are considered susceptible providing the proper environment is present regardless of the level of minor impurities. The literature in cracking of "pure" alloys is summarized by Barnartt (6).

4. Effect of Second Phase Ferrite in Austenitic Stainless Steels

Fontana et al (7) have demonstrated that increasing the ferrite content of austenitic stainless steels increases the resistance of these alloys to stress corrosion cracking in sodium chloride solutions. These data are summarized in Figure 6.

Figure 6 Effect of Ferrite on stress required to induce stress corrosion cracking in several cast stainless alloys. Type 304 and 316 with zero ferrite also plotted. Specimens exposed 8 hours in condensate from 875 ppm chloride water at 400°F.

On the other hand it has been shown that (8) the ferritic alloy, Croloy, will crack as shown in Figure 7. Thus the influence of solution composition must be considered in assessing the value of utilizing second phase ferrite as a crack stopper.

5. Anomalous Cracking in High Nickel Alloys

Based on the evidence given in Figure 5 showing the influence of increased nickel contents in decreasing cracking, it would appear that going to higher nickel contents might offer an easy solution for circumventing stress corrosion cracking. However, two researches have demonstrated that this may not be entirely the case. In one when Ni-O-Nel and Incoloy were exposed to 100 ppm chloride, 50 ppm oxygen at pH 2.8 and 300°C cracks were shown to occur within 500 hours or less. Cracking of the Ni-O-Nel is shown in Figure 8 (8).

Figure 7 Cracking in Croloy 16-1 after 200 hours in water containing 100 ppm chloride, 50 ppm oxygen, at pH 6.5 and 300°C. 85x

Figure 8 Cracking in Ni-O-Nel after 300 hours in 100 ppm chloride, 50 ppm oxygen, at pH 2.8 and 300°C. 75x

In a second case Inconel-600 and several other high nickel alloys have been made to crack in nominally pure water by Coriou (9) in France. However, the inability to reproduce his results in this country have raised questions concerning the possibility of contaminants in the solution. The important point is that for some reason the Inconel did crack in relatively pure water.

6. Effect of Impurities in the Metal

Lang (10) has conducted a study of the effects of impurities on the susceptibility to stress corrosion cracking in magnesium chloride of an 18 Cr - 20 Ni - bal Fe alloy. Additions of nitrogen, phosphorus, arsenic, antimony, bismuth, ruthenium, and aluminum were found to cause cracking in 100 percent of the specimens in a short time. Alloys containing 0.1 percent carbon or 2 percent silicon and no other additions were completely resistant after 30 days and longer. It is interesting to note that the elements in Group V were uniformly detrimental. Also the austenite stabilizers, carbon and nitrogen, had diverse effects.

D. Metal Structure

1. Cold Work

The general effect of cold work is summarized in Figure 9 from work by Hawkes et al (11).

Figure 9 Applied stress vs time-to-failure for 0.020 inch diameter Type 316 wire by boiling 42 percent $MgCl_2$ solution at 154°C.

2. Second Phase

In the case of the martensitic stainless steels, heat treatment has been shown to have a pronounced effect on the susceptibility to stress corrosion cracking. An excellent summary of the literature and an extensive experimental program is presented by Suss (12). In summary, when Type 410 is tempered

in the 650-1100°F range it is susceptible to cracking. Above 1125°F the susceptibility no longer exists. It appears that the presence of inclusions, mainly MnS, is an important factor in initiation of cracking.

E. Environment and Electrochemistry

1. The Oxygen-Chloride Relation for Cracking of Stainless Steels

The general relationship between oxygen and chloride for producing cracking in austenitic stainless steels is shown in the curve (13) below in Figure 10.

Figure 10 Proposed relationship between chloride and oxygen content of alkaline-phosphate treated boiler water, and susceptibility to stress corrosion cracking of austenitic stainless steel exposed to the steam phase with intermittent wetting

The exact role of both the chloride and oxygen is not clear. The oxygen could act either as a depolarizer or as an agent to strengthen the oxide film and enhance selective attack. The precise role of the chloride ions is much less clear.

2. The Relationship Between Cracking and the Pourbaix Diagrams

One of the most significant contributions to the understanding of the role of solution is presented in recent work of Mattsson (14,15).

This work considered the cracking of brass in ammoniacal copper sulfate solutions. Time to cracking vs pH is shown in Figure 11.

Figure 11 Time to cracking of brass (62.8% Cu and 37.2% Zn) as a function of pH in ammoniacal copper sulfate

This time to cracking curve is to be compared with the potential-pH diagram for copper in a water solution containing ammonia as shown in Figure 12.

Figure 12 Potential/pH diagram showing regions for different states of dissolved copper in a water solution with 1.0 g-mol/l of ammonia and 0.05 g-mol/l of copper 25°C

It is readily seen that the onset of cracking in Figure 11 is directly related to the pH at which $\text{Cu}(\text{NH}_3)_2^+$ forms and the minimum in cracking time is related to the formation of $\text{Cu}(\text{NH}_3)_4^{++}$. The variation in cracking pattern with change in pH is shown in Figure 13 below.

Figure 13 Crack appearance in different pH ranges at 140x for brass in ammoniacal copper sulfate solutions

3. Dry Salt

In addition to the cracking observed in aqueous media, dry salt (NaCl) at high temperature ($1100-1400^\circ\text{F}$) has been observed to attack stainless steels intergranularly (16). This phenomenon is aggravated in the presence of stress. While the significance of these data is not completely clear, they should be considered in assessing the role of the chloride ion in stress corrosion cracking

4. Cathodic Polarization

Priest (17) has shown that the application of a cathodic current will stop the progress of stress corrosion cracking in a magnesium base alloy. In the case of stress corrosion cracking in steel Parkins (18) has shown that increasing the stress requires a higher level of cathodic current to prevent cracking. These data are summarized in Figure 14.

Romanov (19) has shown that the cathodic current to prevent cracking in a magnesium base alloy increases with increasing temperature as summarized in Figure 15.

Figure 14 Influence of cathodic polarization on the rate of stress corrosion cracking of steel in boiling solutions of calcium and ammonium nitrate. Initial stresses: 1-43.2 kg/mm²; 2-32 kg/mm²

Figure 15 The influence of temperature on the effect of cathodic and anodic polarization, during stress corrosion cracking of a magnesium base alloy in a 0.1N solution of H₂SO₄ + 35 g/l NaCl

5. Hydrogen Evolution

Priest (17) has shown in the case of a magnesium base alloy that hydrogen evolution accompanies both intergranular and transgranular cracking

6. Specimen Potential During Stress Corrosion Cracking

Numerous studies of corrosion potential of specimens undergoing stress corrosion cracking have been made. Some of the most significant studies have been conducted by Barnartt and van Rooyen (20) and Hoar and Hines (21). Some of the important factors in the composition range of 18-8 stainless steels and

the behavior of elemental chromium, nickel, and iron are summarized in Figure 16 (20).

Figure 16 Corrosion potential vs time for unstressed metals in aqueous 42% MgCl_2 , 146°C, nitrogen atmosphere

The chromium and iron are seen to be much more active than nickel. Such a tendency may be important in the idea advanced by Forty and discussed later in this paper that a tendency to dezincification type phenomena may enhance susceptibility to stress corrosion cracking.

A second important observation appears in the case of the 18-8 composition. It exhibits a sharp decrease in corrosion potential in the unstressed state as a function of time. Note also that the peak is much more pronounced in the unstressed state than one for an alloy not exhibiting stress corrosion cracking in van Rooyan's solutions (the 16-20 alloy).

Cracking when it occurs happens after the peak is reached. It appears that such behavior may be explained by the necessity to form a sufficiently strong oxide film so that the corrodability of unprotected areas would be accentuated and then be more likely to crack. The effect of applied stress has been excellently summarized by Hoar and Hines (21) in Figure 17.

The rise of potential before cracking can occur as shown in Figure 17 and 16 may be an important clue in considering the initiation of stress corrosion cracking.

Figure 17 Typical potential/time curves for various applied stresses. Steel 18-8-Ti, 151-154°C.

7. Temperature of Environment

The general effect of temperature on stress corrosion cracking is demonstrated in Figure 18 (33)

Figure 18 Effect of temperature on time for crack initiation in Types 316 and 347 stainless steels in water containing 875 ppm NaCl

In addition stress corrosion cracking has been observed to occur at room temperature. The temperature at which stress corrosion cracking occurs appears to be sensitive to solution concentration, length of exposure, oxygen concentration, crevices, and chemistry of corrodent.

8. Effect of Various Cations

In the case of chloride cracking of Type 304 stainless steel the type of cation has a strong effect on the time to cracking. As a general rule the metal will still crack and only the rate of cracking will change with the cation. Data developed by Warren (34) shows that Mg^{2+} and Fe^{3+} are the most aggressive with Ca^{++} and Na^+ being progressively less potent. These effects are summarized in Figure 19.

Figure 19 Effect of metallic cation on the time for cracking of Type 304 stainless steel specimens exposed at 100°C to water containing 100 ppm chloride

This general effect has been observed by other investigators

F. Sites of Initiation of Cracking

This section will discuss possible mechanism by which stress corrosion cracking may actually get started or be initiated.

1. Nielsen's Tunnels (22)

One of the strongest evidences for localized attack of the type which could lead to stress corrosion cracking has been uncovered by Nielsen and is shown in Figure 20.

Figure 20 Platinum decoration of reactive sites in annealed Type 304 stainless steel. 8000x

Nielsen used an oxide replica extraction technique to obtain the photograph shown in Figure 20. These tunnels are considered to lie in 110 directions in the 111 plane of the fcc lattice. This appears to be justified further by the branches coming off at 60°.

The implication of this observation is very important from the standpoint of stress corrosion cracking. The 110 direction is the axis of the Cottrell-Lomer type barriers which are hypothesized to exist and be responsible for pile ups within grains. It is likely that the strain energy at these barriers is very high and so it is not unreasonable to suppose that they would corrode preferentially in a solution as aggressive as MgCl_2 to which the above specimens were exposed.

2. Formation of Fans and Trenches

The formation of microscopic fans upon the application of stress and in the presence of an oxidizing atmosphere has been observed in Type 304 stainless steel by Gulbransen (35). The effect of stress on the formation of these fans is illustrated in Figure 21.

The formation of these fans is promoted by the presence of water vapor. It is possible that the trenching which results from the loss of material which makes up the oxide fan may be directly related to the same local reaction involved in the initiation of stress corrosion cracking. A suggested analogy between the fans formed in oxidation and possible electrochemical trenching is shown in Figure 22.

Figure 21 Effect of tensile stress on formation of oxide film formed on 304 stainless steel. Oxidized 24 hours 0.1 atm of oxygen at 600°C. a) No applied stress. 35,000x b) 44,000 psi 9,000x

Figure 22 Analogy between formation of fans and possible electrochemical trenching

This observation is important in that, like Nielsen's, it offers a possible key to explain the initiation of transgranular cracking.

3. Swann and Pickering's Tunnels

Another manifestation of tunneling has been observed by Swann and Pickering (23) in "pre-post" electron microscope studies

of copper gold alloys. In these studies they examined the microstructures of alloys before and after corrosion of thin foils in solutions which were known to cause stress corrosion cracking and in solutions known to be "non-cracking" solutions. Solutions which cause cracking were observed to "drill" holes through the foils. Electron micrographs at progressively longer exposures are shown in Figure 23 together with the hypothesized geometry of the hole.

Figure 23 Growth of pitting corrosion on a Cu-5at% Au alloy exposed to dilute aqua regia. The white angular spots are the pits which have penetrated the foil thickness. Each pit is approximately 40 atoms in diameter. a) (110) face exposed for 10 seconds, b) 20 seconds, c) 60 seconds. d) Schematic representation of corrosion pits as deduced from selected area diffraction patterns and electron micrographs

Solutions which did not cause stress corrosion cracking were found to result in relatively uniform attack with a "hill-valley" structure as shown in Figure 24.

Figure 24 Copper 5 a/o Au exposed to potassium cyanide solution for 30 seconds. Surface roughening is revealed by extinction contours which appear as closed black lines around depressions.

The effect of alloy content and cracking and non-cracking solutions is summarized in the figure below:

Figure 25 Cross sections showing the geometry of attack for Cu-5, 25, and 40 a/o Au alloys exposed to cracking solutions, and for Cu-Au alloys exposed to noncracking solutions. The radius to depth ration of the pits is drawn to scale.

4. Bakish and Robertson's Tunnels

Robertson and Bakish have conducted experiments on Cu_3Au single crystals which were immersed for three months in a 20% FeCl_3 solution. Results of metallographic examinations are shown below. (32)

Figure 26 Cu_3Au single crystal strained 5% and immersed for three months in 20% FeCl_3 . a) Crystallographic planes represented by the micrographs; b) chemically reactive sites lying in traces of slip clusters on the external surfaces; c) a cross section parallel with the slip plane showing reaction along the 110 direction; d) traces of slip lines from deformation after sectioning the crystal.

In this work as in that of Nielsen's cracking proceeds in the 110 direction in 111 planes. The evidence presented in general appears to be strong that local tunneling reactions may be a very important factor to consider in stress corrosion.

5. Local Chemical Inhomogeneities Due to Coarse Slip

Swann and Pickering (24) suggest that the initiation of sufficiently localized corrosion to lead to stress corrosion cracking may be related to chemical inhomogeneities which are a direct consequence of coarse slip. In the case of coarse slip a large number of dislocations come to the surface on one slip plane. If it is hypothesized that these dislocations may drag with them solute atoms, then a consequence of the coarse slip is that these solute atoms would be deposited on the metal surface as the dislocations emerge. If the regions where dislocations come to the surface are widely separated as would be the case in coarse slip, then widely separated chemical inhomogeneities would result. These would serve as anodic regions of the long and narrow type and serve to initiate cracking. This phenomenon may be related to that hypothesized by Gulbransen in the case of fans in the discussion above.

The conditions responsible for coarse slip are those tending toward restricted cross slip (to be discussed later); i.e. if dislocations cannot cross slip, they all tend to stay in a single plane. If dislocations were able to cross slip, solute atoms would be distributed more uniformly and the possibility of intense corrosion at a few areas would be much reduced. The significance of coarse and fine slip is depicted schematically in Figure 27.

(a) FINE SLIP, CORRESPONDING TO A CELLULAR DISTRIBUTION OF DISLOCATION TANGLES

(b) COARSE SLIP, CORRESPONDING TO A PLANAR ARRANGEMENT OF DISLOCATIONS

Figure 27 The effect of slip behavior on the number of "fresh" atoms exposed at the surface of a metal during deformation

Swann and Pickering consider that the tendency toward bringing

atoms to the surface can be related to strain aging observations. This is based on the assumption that strain aging is due to diffusion of solute atoms to dislocations. This in turn pins this dislocation and prevents its movement until an extra increment of stress, $\Delta\sigma$, is applied. This results in a yield point or $\Delta\sigma$ effect.

Once a crack is initiated at a local region, it would continue to have a fresh supply of solute atoms resulting from the movement of dislocations to the crack tip.

6. Reactivity of Stacking Faults

Swann and Nutting (25) have observed that stacking faults in thin films can be preferentially attacked in copper aluminum alloys. Swann and Nutting (26) also observe that upon aging solute atoms will diffuse to stacking faults. Such diffusion may be the cause or at least may enhance the reactivity of stacking faults. Thus preferential chemical attack along a stacking fault may be another possibility for reaction which is sufficiently localized to initiate stress corrosion cracking.

7. Reactivity as Indicated by Etch Pitting

Numerous experiments have demonstrated the chemical reactivity of dislocations in the form of etch pits at metal surfaces. Evidence that the strain energy alone is sufficient as the driving force for dissolution is provided by Seastrom (27) who has conducted etch pitting experiments on iron whiskers. In the case of a pile up of dislocations at a barrier, there would be a series of etch pits formed. The series of etch pits is again similar to the type of localized chemical attack which might lead to initiation of stress corrosion cracking. The same factors would promote initiation of this type, i.e. restricted cross slip, as are considered to promote the formation of tunnels, fans, and coarse slip.

8. Anodic Regions Due to Hydrides

Vaughn et al have made extensive studies into the role of hydrogen in stress corrosion cracking of Type 304 stainless steels. They consider that cracks may be initiated by the anodic dissolution of a chemically active hydride phase. They present evidence that a hydride exists in the iron-chromium-nickel alloys system which is anodic to its surroundings. (28) Consideration of the propagation of cracking due to the influence of hydrides will be considered later.

G. Role of Stacking Faults and Dislocations in Stress Corrosion Cracking

Since some of what has been discussed above and much of what is to follow is related to stacking faults and dislocations, it is

appropriate to stop here to clear up a few points.

It is well established that the existence of dislocations is required for plastic flow. There are two types of dislocations: edge and screw. Often a given dislocation line may be a combination of both types.

When sufficient stress is applied to a crystal, dislocations move and only in certain planes and in certain directions. When a moving dislocation encounters a barrier (the origin of this barrier will be discussed shortly), it is stopped and cannot move further unless it can in some way move around the barrier. The screw dislocation can move over a barrier by "cross-slip" but the edge cannot. (Cross slip is a change in plane of movement from one plane in a given crystal system to another which is usually inclined at a low angle in the direction of movement.) This inability of a edge dislocation to cross slip is based on the relative directions of the movement of the line and the direction of the dislocation line itself. If a given dislocation has both edge and screw character, it cannot easily cross slip.

In connection with the relative ease of cross slip is a very important quantity called "stacking fault energy." Certain dislocations have the property of being able to dissociate into two partial dislocations. These two dislocations must always have partly edge and partly screw character. This is dictated by the geometry involved and is true even if the original dislocation were 100% of the screw type. It should be kept in mind here that the edge component will resist cross slip around barriers. After the dislocation has dissociated into two partials, there is a region between the two which is known as a "stacking fault." This results from the fact that while outside the two partials the crystal may have perfect packing, say ABCABCABC . . . the atoms in the faulted region have ABCABABCABC . . . packing.

The stacking fault energy is the energy required per unit area of the faulted region. If this energy is very high, then the distance between partials may be very small or possibly the original dislocation may not even dissociate. If the stacking fault energy is low, then the partial dislocations will be widely separated and the probability of single dislocations dissociating is increased. Thus, as the stacking fault energy decreases, the probability of forming partial dislocations increases; this in turn will give edge character to screw dislocations; which in turn reduces the possibility of cross slip.

The reduced possibility of cross slip has several important implications:

1. The number of dislocations piled up behind a barrier will be relatively high and the associated strain energy will be high. There is also an increased possibility that solute atoms will diffuse toward these regions.

2. The tendency for slip to be confined to widely separated slip planes is increased.
3. The possibility of having the type of chemical inhomogeneities shown in Figure 27 is increased.

A high stacking fault energy will lead to the reverse of the above; namely, cross slip and a more uniform and less concentrated distribution of solute atoms on the metal surface.

Several other points regarding stacking faults and stacking fault energy are worth noting at this point as follows:

1. Barriers within crystals are hypothesized to form from the interlocking of stacking faults from different planes. The strength of these barriers, i.e. the number of dislocations which they can cause to pile up without breaking down the barrier, increases with increasing extension of the stacking fault. Thus, low stacking fault energy favors wide extensions which favors strong barriers.
2. Lower stacking fault energy should favor stress corrosion cracking since it reduces or restricts cross slip, makes strong barriers, permits the formation of the stacking faults with their associated chemical reactivity, and promotes local chemical inhomogeneities on the metal surface.

Evidence for the relationship between lowered stacking fault energy and increased susceptibility to stress corrosion cracking comes from a number of sources. Significant ones of these are discussed below.

In an article by Swann (29) the susceptibility to transgranular cracking of copper-zinc and copper-aluminum alloys is related to stacking fault energy, alloy composition, and the pattern of dislocations as observed in electron transmission. A graph depicting these relationships is shown in Figure 28.

It is seen immediately in Figure 28 that alloy composition has an important effect on stacking fault energy. There appears to be critical stacking fault energy below which transgranular cracking will occur in preference to intergranular cracking. It is difficult to determine whether this affects the formation of barriers, the nature of cross slip, or some other factor.

Electron transmission photomicrographs shown in Figure 29 illustrate the difference between the structure of a pure metal not susceptible to cracking and an alloy susceptible to transgranular cracking. In the case of the copper the stacking fault energy is

Figure 28 The relationship between the stacking fault energy, γ , (in units of 10^{-3} Gb), dislocation distribution, and the mode of observed stress corrosion failure of Cu-Zn and Cu-Al alloys. Approximate composition limits for transgranular cracking.

Figure 29 On the left is pure copper elongated 8 percent showing the cellular distribution of dislocation tangles typical of deformed pure metals. On the right is Cu-8w/o aluminum elongated 5 percent, showing dislocation arrangement typical of low stacking fault energy metals.

high and cross slip is favored. As a result of extensive cross slip, tangles are formed as shown in the left hand picture in Figure 29. Since cross slip is not favored in metals with lower stacking fault energy, the dislocations tend to stay in planar arrays as shown in the case of the right hand picture.

Now it is possible to rationalize the effect of nickel on the stress corrosion cracking of stainless steels as shown in Figure 5. Figure 30 shown below relates time to failure, nickel content, stacking fault energy for variations in nickel content.

Figure 30 The effect of nickel on the stacking fault energies and times for stress corrosion failure of approximately 18 percent Cr stainless steels.

Again, in the case of the above figure, stress corrosion susceptibility appears to be related to a decrease in stacking fault energy. To illustrate the way in which both nitrogen and nickel are related to dislocation distribution and stress corrosion susceptibility Figure 31 is included. The two photographs on top have low nitrogen and are not susceptible to cracking. Note the cellular structure. The two alloys on the bottom have higher

Figure 31 a) and c) have approximately 0.004N and b) and d) have approximately 0.1N. a) and b) have approximately 12.8 Ni and c) and d) have approximately 23 Ni.

nitrogen and consequently have planar dislocation arrays. These alloys are susceptible to cracking. The alloys on the left have low nickel and the ones on the right have high nickel. There appears to be corresponding increase in degree of cellular structure with increasing nickel content.

A further evidence of the intimate relationship between stacking fault energy and stress corrosion susceptibility comes from studies of twin frequency and stress corrosion cracking susceptibility. Since a twin is also a stacking fault, ... ABCABCACBACBA ... but with a lower energy requirement due to being equivalent to one half the perturbation involved in a stacking fault, the twin frequency should give a good indication of stress corrosion cracking susceptibility. Such a plot of twin frequency and stress corrosion cracking susceptibility is shown in Figure 32 (30).

Additional evidence for this relationship comes from Barnartt and vanRooyan where twin frequency is related to experimentally measured stacking fault energy in the table on the following page.

Figure 32 Comparison of dependence of stacking fault energy (twins per grain) and transgranular failure of approximately 17% chromium stainless steel on nickel concentration. (Cracking data from Copson).

Table II

Measurements of the Number of Twins per Grain and Stacking Fault Energy

Base Material: 16%Cr-20%Ni steel			
Addition, %	Mean Grain Size	Mean Number Twins/Grain	Mean stacking fault energy (erg/cm ²)
none	5.0	2.2	+25
1.45 Mn	6.0	3.0	+25
0.54 Ti	5.0	4.0	20
1.51 Mo	5.5	4.3	15

III CURRENT THEORIES OF STRESS CORROSION CRACKING

After the above review of data it is readily seen that some very specific guesses can be made concerning mechanisms of stress corrosion cracking.

Several specific overall theories of stress corrosion cracking have recently been proposed. At certain points these theories advance hypotheses which are yet to be evaluated experimentally. However, much of the data cited previously does support portions of the theories described below.

The various theories are discussed only briefly. Some of the elaborate detail involved is better understood only by reading the original papers,

A. Forty's Theory (31)

Forty's theory for stress corrosion cracking is based on his observation in brass in which he saw that cracks propagate in short bursts and are arrested at pre-existing slip bands. The factor determining the overall rate of cracking appears to be related to chemical dissolution while the crack was held up at a slip band. Chemical attack appears to move parallel to the slip band. When sufficient chemical dissolution had weakened and embrittled the structure, the crack would progress again in a short burst. Schematically, the progress of cracking is shown in figure 33.

Figure 33 Stages in cracking according to Forty's hypothesis

Based on the above observations Forty has outlined the following theory of stress corrosion cracking:

1. Cracking is initiated as a result of chemical processes leading to surface defects which are of a size and geometry to concentrate stress. These defects may result from preferential attack along stacking faults during the corrosion stage at a pre-existing slip band. The initiation stage at the metal surface may result from a dezincification phenomenon in which the more active zinc atoms stay in solution and the copper redeposits. The redeposited structure is full of defects which under the influence of stress and thermal activation tend to coalesce into pores.
2. Pores in a plane of slip will lead to concentration of stress and give an increased normal stress according to the diagram in Figure 34.

Figure 34 Showing distribution of normal and shear stresses around an elliptical hollow dislocation

Dislocations continue to pass into the above cavity until the hole reaches a critical size after which fracture occurs. Fracture will propagate in a brittle manner.

3. Forty's general assumption is that the fracture between slip bands is essentially a brittle mechanical fracture. He rationalizes the possibility of brittle fracture in a normally ductile material by invoking various considerations relating to solid solution hardening.

B. Robertson's Theory for a Unified Mechanism

Robertson (32) has proposed a unified theory for the cracking by both transgranular and intergranular modes. His theory requires a) a susceptible path and b) a barrier against which dislocations can pile up to exert sufficient stress to crack the chemically weakened path.

A susceptible path in intergranular cracking is considered to be the chemically active grain boundaries or anodic precipitates which may form preferentially at grain boundaries. In the case of transgranular cracking a susceptible path for chemical attack may be provided by one or several of the mechanisms described in section II F. The barriers for pile up of dislocations in intergranular cracking are grain boundaries and in transgranular cracking are probably of the Cottrell-Lomer type.

The choice of a particular metal for inter or transgranular cracking is related to the relative corrosion potentials of tunnels and grain boundaries together with the relative strengths of barriers.

The corrosion potential will be affected by such factors as pH, ionic composition, temperature, metal alloy and impurity composition, and metal structure.

A schematic illustration of a dislocation pile up at a grain boundary barrier is shown in Figure 35.

Figure 35 Maximum normal stress, σ_m , at the head of a group of dislocations piled up at a grain boundary, oriented with respect to the slip plane so that the plane of maximum normal stress coincides with the boundary plane.

Applying this to a multigrain system we have Figure 36.

Figure 36 A model for intergranular corrosion cracking involving stress concentration at reacting grain boundaries followed by fracture due to the component of normal stress operating across a boundary plane.

The sequence of events indicated in the above intergranular cracking is as follows:

1. Plastic flow results from crevices due to preferential grain boundary attack.

2. Dislocation pile ups due to plastic flow.
3. Chemical attack proceeds and makes cavity.
4. Stress breaks weakened bonds and dislocations run into chemically formed cavity.

In the case of transgranular cracking the above sequence also holds except that the location and detailed nature of the dislocation pile up are somewhat different. Illustrated in Figure 37 is a dislocation pile up and crack nucleus at a Cottrell-Lomer type barrier.

Figure 37 Crystal geometry of the Cottrell-Lomer barrier and its relation to crack nucleus. Note the accumulations of dislocations on two intersecting slip planes.

In his paper, Robertson gives the detailed criteria and general experimental support for his proposed mechanism.

C. Vaughn's Hydrogen Theory (28)

The essential features of this theory are as follows:

1. Cracking is initiated at alloy hydrides on the metal surface which have corroded preferentially due to their anodic nature.
2. Certain alloys are susceptible to cracking because on the application of stress the hydrides line up in general perpendicular to the applied stress. Alloys which are not susceptible do not experience this realignment.
3. Cracking proceeds by the combined action of chemical dissolution of the anodic hydride phases and the effect of mechanical stress on the brittle hydride phase.
4. As a result of anodic dissolution, there is no trace of the initial alloy hydride.

Photographs in Figure 38 show the difference between a susceptible and non-susceptible alloy. The main difference appears to be in the line up of precipitate in the susceptible alloy and the general randomness in the non-susceptible material.

Figure 38 Photomicrographs of annealed Type 304 stainless steel. Alloy on left is susceptible and was stressed in tension in the vertical direction. Alloy at right received similar treatment but is not susceptible. 250x

Figure 39 shows the microstructure of cracks in a susceptible alloy containing hydride platelets.

Figure 39 Photomicrograph of stress corrosion crack in susceptible Type 304. U-bend, hydrided. 240x

IV. CONCLUSIONS

There is an abundance of experimental information which together give a much improved understanding of stress corrosion cracking. It appears that a single "best" theory may be related to considerations of a) the potential-time curves of Hoar and Hines, b) the Pourbaix diagram suggestion of Mattsson, c) the tunneling observed by several investigators, d) influence of stacking fault energy and dislocation distribution advanced mainly by Swann, and e) the unified model of Robertson's.

It appears that one of the important conclusions is that testing should be conducted on an integrated basis which considers electrochemistry, physical metallurgy, and stress corrosion cracking studies. While considering the integration of these three approaches, the simultaneous use of several alloy compositions appears to be important. It appears that tests based on the $MgCl_2$ test are in themselves no longer adequate for studying the mechanism of cracking.

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