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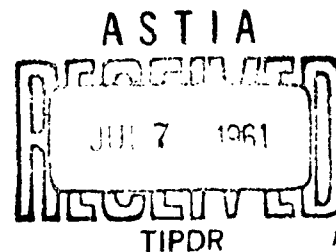
STRESS CORROSION CRACKING

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**SOME ASPECTS OF STRESS CORROSION CRACKING IN
AUSTENITIC STAINLESS STEELS**

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August 25, 1960

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ABSTRACT

The addition of small amounts of nitrogen, or larger amounts of molybdenum, reduces the resistance of austenitic stainless steels to transgranular stress corrosion cracking. Increasing the carbon or nickel contents, and perhaps tin to some extent, makes the alloys more resistant to cracking. These variations in time to failure result from differences in the reaction time preceding crack propagation, while the rate of crack penetration remains approximately constant.

Composite stress corrosion samples were made in which the test alloys were bonded to a steel which was susceptible to cracking. Results showed that there is good correlation between the times to failure of the experimental alloys when tested alone, and the number of propagating cracks which will penetrate into them. The possibility is suggested that crack initiation consists of a form of corrosion attack which is required to expose a path of rapid corrosion, after which true cracking proceeds.

Potential measurements indicate that strain-induced anodic depolarization cannot account for cracking entirely by an electrochemical mechanism. Preference is given to an electrochemical mechanism wherein reaction takes place at sites of structural and composition changes in the steels. These may be formed under the influence of stress, setting up paths of easy localized corrosion.

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SOME ASPECTS OF STRESS CORROSION CRACKING IN AUSTENITIC STAINLESS STEELS

by

D. van Rooyen*

INTRODUCTION AND SUMMARY

Most of the reported work on the effect of composition upon the transgranular stress corrosion cracking properties of austenitic stainless steels concerns itself with the effect of nickel on this type of failure. Rocha⁽¹⁾ showed that an increase in nickel content causes improved resistance to cracking. Edeleanu⁽²⁾ tested some very highly resistant steels with nickel contents between 16 and 21%, some of them remaining unbroken in boiling 42% magnesium chloride for 500 hours at stress levels of 40,000 psi. Scheil⁽³⁾ and Copson⁽⁴⁾ reported the onset of immunity to stress corrosion cracking at 45% or more nickel, while Edeleanu and Snowden⁽⁵⁾ pointed out the superiority of stainless steels with 30% nickel over those with lower nickel contents.

Most of the results on the effect of additions other than nickel appear in the work of Uhlig and White⁽⁶⁾, who used steels of low (8%) and high (20%) nickel contents as basic materials for study: In the case of α -phase 8% nickel alloys, which do not crack, the addition of carbon, nitrogen, or niobium produces sensitivity to stress corrosion cracking, with titanium having no effect. The addition of titanium, or niobium, to austenitic stainless steel of the 18-8 type has little or no effect, all the alloys fail in a short time. Stable 20% chromium-20% nickel austenitic steel alloys are resistant to cracking at low nitrogen levels, i.e., around 0.002% N_2 . Higher nitrogen contents in these materials cause cracking, while increasing carbon contents increase the resistance to failure. Silicon

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increases the resistance to cracking of 18-8 and 25-20 Cr-Ni type steels, with cobalt and probably boron also being beneficial. The observation of Uhlig and White on the effect of silicon is in conflict with the results of Leu and Helle⁽⁷⁾, who found that the number of observed cracks is increased by the addition of silicon.

Attempts to explain the mechanism of cracking and to show how this is changed by various alloy additions are described in the literature. In the 18-8 series of alloys, cracking is prevented by those factors which give rise to a predominantly ferritic structure, since it seems that transgranular stress corrosion cracking in chloride solutions is confined to the steels with a face-centered-cubic structure. For 18-8 type steels containing enough carbon and nitrogen, the structure is austenitic, but some martensite can be formed either by sufficient cooling of the alloy or by straining it at a temperature below a certain critical value. Edeleanu⁽⁸⁾ showed that stress corrosion is more rapid when some martensite is present, but it seems to be generally believed that such a phase change is not always the cause of cracking.^(6,9,10) Uhlig^(6,11) explains the mechanism of cracking as a process which is partly electrochemical and partly physical, the crack sensitive paths being associated with plastic deformation and localized composition gradients; the adverse effect of nitrogen is assumed to be associated with its diffusion to lattice imperfection sites, forming cathodic areas; the beneficial effect of silicon and nickel, on the other hand, is ascribed to their "altering the pattern of cathodic atmospheres or precipitates, or inhibiting their formation." Hoar and West⁽¹²⁾ support the electrochemical mechanism of cracking, and showed that pure nickel, which does not crack, has

corrosion characteristics different from those of a susceptible stainless steel. Their tests involved the measurement of polarization of specimens which were corroding at high current densities, and they indicate that nickel does not undergo stress corrosion cracking because excessive polarization would prevent cracks from penetrating by an electrochemical corrosion mechanism. By interpolation, it would be expected that the effect of increasing nickel contents in stainless steel would be one of making rapid corrosion more and more difficult, until a point is reached where crack propagation is no longer possible.

In the work reported here, two series of iron-base alloys were examined: 16% chromium-14% nickel and 16% chromium-20% nickel. In both materials the addition of carbon is beneficial. Additions of silicon, nitrogen, manganese, titanium, aluminum, and combinations of the metallic additions with nitrogen, have no significant effect on the times to failure of the 14% nickel alloy. This alloy, without any additions, is not very resistant to cracking. The effect of platinum is to lower the time to failure by an order of magnitude. The "pure" 20% nickel alloy is very resistant to cracking and never fails within the test periods. The addition of manganese to this resistant material has no effect, while nitrogen, or molybdenum, causes it to become susceptible. The adverse effects of nitrogen and molybdenum can be offset by carbon additions. The simultaneous presence of manganese, silicon, and molybdenum in several alloys with 20% nickel results in short times to failure, while similar compositions without molybdenum do not fail within the test periods. Three nickel-base alloys were tested and found to be immune to stress corrosion cracking in magnesium chloride.

As a result of the knowledge gained during the early stages of the investigation, four alloys were selected for further tests. They have the same nickel and chromium contents, but times to failure are within the range of 12 to more than 350 hours. The resistance of these alloys to stress corrosion crack propagation was evaluated with compound specimens, prepared with each of the test materials bonded to a susceptible commercial steel. Cracks were started in the commercial alloy, and their behavior was observed when they reached the boundary between the two metals. This technique eliminated the initiation period for the test alloys.

Cracks penetrate readily into the steel which has a short life when tested alone, but such crack penetration into the alloys with longer times to failure becomes less frequent. The most resistant alloy does not support any cracking. The results show that the number of areas which are immune to cracking increases with increasing time to failure, suggesting that the time to failure is dependent on the ease with which a susceptible site is exposed during the incubation period.

Studies were made of the electrochemical behavior of rapidly corroding and rapidly yielding alloys, in the absence of concentration polarization. There is no correlation between the strain-induced anodic depolarization and the degree of susceptibility of the alloys to stress corrosion cracking. Also, when some depolarization is found, it occurs only at current densities of about one order of magnitude lower than that which would account electrochemically for the observed rates of crack propagation. While there is no disagreement with the electrochemical theory of stress corrosion cracking, it is suggested that rapid localized corrosion does not mainly result from the lack of polarization at the

roots of cracks; this interpretation is different from that of Hoar and West, who reported similar work in 1958. Results tend to favor suggestions in the literature which propose composition gradients within those alloys which suffer stress corrosion cracking. These changes could possibly be caused by straining, and may account for the necessary driving force required to sustain rapid corrosion along crack-sensitive paths.

EXPERIMENTAL PROCEDURE

The 304 type stainless steel* was in the form of cold-worked wire. The alloys containing 14% nickel were made by levitation melting samples of approximately 15 to 20 grams each in an argon atmosphere; the starting material was a vacuum-melted "pure" alloy of composition 16% chromium, 14% nickel. The steels containing 20% nickel were taken from 1/16 in. diameter wire samples which were prepared by vacuum-melting and subsequent reduction. The reason for selecting 16% chromium-20% nickel and 16% chromium-14% nickel alloys as basic compositions was to have available two stable austenitic alloys, even with low carbon and nitrogen levels. Where additions were made to these compositions, specimens were obtained from larger pieces of vacuum-melted material, and the new alloys were prepared by the levitation technique. The preparation of wires of 0.025 in. diameter was done in successive steps of 50% reduction and vacuum annealing at about 1065°C. All stainless steel samples were tested after a final vacuum anneal at 1065°C $\pm$ 5°C, followed by cooling in vacuum. The temperature fell to

*Kindly supplied by Driver-Harris Company.

below 400°C within about 20 seconds. The nickel-base alloys were vacuum-annealed to give an expected grain size of 5 to 6.

The chemical compositions of the alloys are given in Tables I, III, V, VII, and IX.

The wires were tested in boiling 42% magnesium chloride solutions in horizontal glass cells, the tensile stresses being applied by direct loads, as described elsewhere.⁽¹³⁾ A metallographic examination was made of a longitudinal section of each tested alloy to confirm that failure, when it occurred, was a result of stress corrosion cracking.

Bonded specimens were prepared by arc-depositing commercial 304 or 347 type stainless steel onto 1/8 in. thick strip of each of the selected test alloys. A piece 6 inches long and 1/4 inch wide was then cut out to include some of both materials. This strip was hot rolled to 0.05 in. in thickness, pickled, and electropolished in a mixture of phosphoric acid and glycerol. The edge of the commercial alloy side of the piece was then machined in such a way that two-thirds of the width of the specimen consisted of the test alloy. All specimens were annealed in vacuum as described for the wire samples. Testing was carried out in glass cells containing boiling 42% magnesium chloride, at stress levels of 27,000 psi. During some of the tests a current of 5 milliamps/cm² was passed between the specimen (anode) and an auxiliary stainless steel cathode immersed in the same solution. This caused rapid and prolific cracking in the commercial material. When the cracks reached the interface between the metals, the stress on the test alloy was 40,500 psi, not taking into account any stress-raising effect at the roots of the cracks. The duration of the tests was from

6 to 9 hours, except where failure occurred before this time. After testing, pieces were cut from the samples, mounted, polished and etched for microscopic examination.

The apparatus for carrying out the depolarization tests on rapidly yielding wire samples is shown in Fig. 1. Potentials were measured against saturated calomel electrodes which were connected to the apparatus at the points marked "salt bridge." The potential difference was recorded on a "Brush" high-speed recorder which was connected to the output terminals of a "Keithley 600" electrometer. The input of the electrometer was connected to the test apparatus. A direct current was impressed between the specimen and the platinum cathodes A and B. The boiling magnesium chloride was forced through the test cell by compressed air, regulated so that the liquid flowed over the specimen at 100 cm/sec. The wire was stretched by means of an electrically-driven pulley at rates of 7 and 12%/sec. In some tests a weight was released to stretch the wire during the test by about 15% in a small fraction of a second, giving a strain rate well in excess of 100%/sec. The cell generally used was made of Pyrex glass with platinum cathodes, Fig. 1; other cells were also used. These were made out of Pyrex glass with full-length stainless steel wire cathodes, or from a copper tube which also acted as a cylindrical cathode around the test wire. In the latter case the tube was preheated to 140°C by an electrical heating element. These two different cell designs contained probes similar to the one shown for potential measurements in Fig. 1. In all cases, except where the cathode did not run the full length of the test wire, results were unaffected by cell design or preheating. The difference caused by using only cathode A

will be discussed later in this paper. The specimens were vacuum-annealed and cleaned in the usual way before testing.

The electrochemical tests were of very short duration, and it would have been better if the testing times were longer. Unfortunately, the strain rates used made this impossible.

RESULTS

(a) Nickel and nickel-base alloys. Samples of nickel and the three nickel-base alloys were found to be immune to cracking in boiling magnesium chloride within the test period of 500 hours, at stress levels of 35,000 psi. Under similar conditions a commercial 18-8 type stainless steel failed in about two hours. The compositions of these materials, and details of the results, are included in Tables IX and X.

(b) Stainless steels with 20% nickel. The materials in this series had stable austenitic structures. The purest alloy was prepared by vacuum-melting, and contained 0.01% carbon, with no other intentional additions. No cracking of this sample was found in any of the tests, which lasted up to 350 hours in some cases. Other alloys, prepared by vacuum-melting, but containing 1.5% manganese, or 0.1% carbon, also did not fail. The alloys containing molybdenum cracked in about 10 to 120 hours, depending on the molybdenum content. Carbon was added to alloys containing small amounts of manganese and molybdenum. The levitation technique was used, in an atmosphere of argon. As the carbon content in the molybdenum alloy increased, improved resistance to stress corrosion

cracking was observed. These results are included in Tables II, IV, VI, and Fig. 2. (The latter curves were drawn through only a few points, and should thus be used for a better general understanding rather than as absolute information). When nitrogen was added to samples of the alloy containing manganese, the times to failure decreased with increasing nitrogen content, and resistance could again be improved by increasing the carbon content.

Several air-melted alloys were examined, the results being in general agreement with those obtained from the vacuum-melted materials. Thus, all samples which contained molybdenum failed in short times. The presence of silicon and manganese did not change the slight resistance to stress corrosion cracking of the alloys containing molybdenum, while samples containing very little (0.002%) molybdenum, together with 0.5% silicon and 1.5% manganese, did not fail within the test period. Variations of the molybdenum content, in combination with varying amounts of manganese, did not give rise to significant changes in results.

(c) Stainless steels with 14% nickel. The materials tested in this series were nonmagnetic. The purest alloy in the group failed in about 30 hours. The addition of various levels of manganese, silicon, aluminum, titanium, or nitrogen caused only minor variations in times to failure, as did combinations of the listed elements with nitrogen. Carbon additions increased the resistance of the alloy to stress corrosion cracking. No failures were found within the test period at around 0.1% and higher carbon levels. In these compositions the simultaneous addition of some titanium, silicon, niobium, and manganese could be tolerated without impairing their resistance. For some levels of tin content

somewhat longer times to failure were found. The addition of 0.1% platinum to the "pure" base alloy reduced the failure time to about 3 hours. Results are shown in Table VIII and Figure 2.

(d) Electrochemical potentials of specimens under static load. The electrochemical potential changes of some of the specimens were made during tests at constant load. These curves are given in Figs. 3, 4, 5, and 6. The general behavior is found to be very much the same as for commercial steels.⁽¹⁴⁾ Initially, the potential is relatively low, and a slow rise takes place during the first hour or more. This period is followed by a steeper rise, after which it remains more or less constant. No further sudden changes in potential are observed in the case of the alloys which do not crack. In the case of susceptible alloys, failure is preceded by a rapid fall in potential. The time between the onset of the rapid fall in potential and final fracture is about the same as in the case of commercial steels, and seems to be independent of the total time to failure. In the case of the steel containing molybdenum, for instance, the duration of the final potential drop was 15 minutes, while it was 20 minutes for the alloy with nitrogen. The total times to fracture of the materials were 18 and 17 hours, respectively. The 18-8 type stainless steel samples had failure times of about 2 hours, with the period of rapid potential fall lasting for 6 to 20 minutes. In the 18-8 type steel, it was shown that cracking does not take place before the final potential drop.^(13,14) Similar tests were made for the present alloys, i.e., the ones containing molybdenum and nitrogen: the alloys were left in the test solution for prolonged times while stressed at 40,000 psi, but removed before final fracture.

No cracks were found in the sections of the wires removed before the onset of the rapid drop in potential, in spite of the many hours since passing the peak in potential, as shown in Figs. 7 and 8. In making the sections of the wires the entire specimen was mounted as shown in Fig. 9, so that a maximum area was exposed by polishing before inspection for cracks.

(e) Bonded samples. Tests were made with nickel-base alloy No. 1, and three special alloys of basic composition 16% chromium-20% nickel: one had no further additions (VM-462), the second contained 1.5% molybdenum (VM-463), and the third contained 0.5% nitrogen (VM-467). Detailed analyses are given in Table I. Cracks did not penetrate into the nickel-base alloy, or the "pure" alloy (VM-462). When corrosion was stimulated by means of an applied direct current, pitting started at the roots of many of the cracks upon reaching the interface between the test alloy and the susceptible material. Cracks that stopped, and also pitting, are shown in Figs. 10 and 11. Figure 12 shows some of the cracks which penetrated into the molybdenum-containing alloy. Similar penetration could be observed over most of the specimen but, at a few sites, some of the cracks did not exceed the boundary line; an example of this is shown in Fig. 13. The nitrogen-containing alloy behaved in a manner exactly opposite to the one with molybdenum: most of the cracks stopped at the boundary, but a few penetrated all the way through the test alloy, which ruptured after 5 hours. For comparison, the times to failure are repeated for the alloys when tested alone: the nickel-base alloy and VM-462 did not break; VM-463 broke in 12 to 17 hours, and VM-467 broke in 125 hours or more.

(f) Anodic polarization during yielding. Upon bringing the flowing liquid into contact with the test specimens, and before applying strain or current, the potential readings were about -400 mv., as measured against a saturated calomel electrode. High anodic current densities between 0.35 and 0.7 amps/cm² caused a rise in potential, of the order of 300 millivolts. This value was unaffected by straining the samples, when readings were made at either point 1 or 2 (Fig. 1) and with both cathodes in operation, or when made at point 1 with either both cathodes or with only cathode A in operation. When the reference electrode was placed at point 2, and only cathode A was connected, a smaller initial rise in potential was observed upon switching the current on. In this case straining caused a lowering of the corrosion potential to a value very close to that of the unpolarized metal. Detailed results are given in Table XI.

Samples of AISI Type 304 steel were used to obtain these preliminary results, which indicated the probability that there was an uneven current distribution over the test wire when cathode B was not operative. Uneven current distributions result in lower current densities over the lower portion of the wire, and this is accentuated by the constriction in the cell. Thus, when the reference electrode is connected to the lower part of the equipment (point 2), the potential difference is largely that between the calomel electrode and the portion of the wire corroding at low current densities. To verify this, tests were carried out in which both cathodes were connected in the circuit, but much lower currents were passed (0.03 amps/cm²). When this was done the initial rise of potential upon starting the current was of the

same order as when high currents, cathode A, and point 2 were involved. Straining lowered the measured potentials to final values that were 40 to 60 millivolts more positive than the unpolarized alloys. No IR drop corrections were made. In all these measurements at the lower current densities the readings at points 1 and 2 were the same. Not all the test alloys showed the same degree of strain-depolarization at low current densities, the values ranging from 5 millivolts for nickel "61" to 85 millivolts for nickel "A." The nickel-base alloy No. 1, type 304 stainless steel, and the three special 20% nickel steels showed intermediate values for the drop in potential during straining. When larger rises in potentials occurred as a result of current flow, depolarization was large and vice versa. In this way the final values obtained were not very different for the various compositions, regardless of their crack susceptibility.

In no case was any difference observed between liquid flow rates of 50 or 100 cm/sec. Also, there was no detectable effect caused by changing the strain rate from 7 to 12%/sec. In the few cases, at high current densities, where strain rates in excess of 100%/sec. were used, this too made no difference in the results.

DISCUSSION

A comparison of the results obtained with the 18-8 type steel, the 14% nickel, and the 20% nickel alloys shows a gradual increase in resistance to stress corrosion cracking as the nickel content increases. This supports the evidence put forward in the literature. At these levels, however, the nickel

content alone does not guarantee any specific degree of resistance, since it also depends on the presence of other additions, such as relatively small amounts of nitrogen or larger amounts of either carbon or molybdenum.

The role of minor additions, which reduce the time to failure of otherwise resistant alloys, is one of reducing the time of crack initiation. Once a crack has started, however, the propagation time is about the same for all the alloys tested. Evidence for this is found in the fact that no cracks have been observed in susceptible specimens after many hours of testing, but before the onset of the final rapid drop in electrochemical potential, which precedes total fracture, as shown in Figs. 2 to 8. The interpretation of these potential curves is analogous to those for commercial alloys^(13,14), wherein it was demonstrated that cracks do not form before the final fast fall in potential, and that the total period of crack propagation was within 6 to 20 minutes. In the present results these times are essentially the same. Two possibilities may explain the increase in times to initiate cracks as the composition of an alloy is changed towards one of greater resistance to cracking: (a) the rate or nature of the initial surface reactions may be influenced by composition, so that the required chemical and geometrical conditions for cracking are attained more and more slowly or, (b) the number of susceptible paths in the alloys may decrease as the composition approaches immunity to cracking, so that longer times of corrosion attack will be necessary before such a susceptible path is exposed to the corrosive medium. Because the time required for cracks to penetrate into the alloys remains about constant, it seems reasonable to assume that there is very little difference in the nature of the crack paths in any of the various susceptible alloys.

The adverse effect of noble metal additions on cracking resistance can be explained by local cell action, in which the steel is the anode. Impressed anodic currents can cause passivity in stainless steel^(15,16), and similar results have been obtained by noble metal additions⁽¹⁷⁾; in the presence of stress, however, stainless steel in boiling magnesium chloride shows more prolific cracking when the anodic reaction is stimulated by impressed currents.⁽¹⁸⁾ Also, coupling to noble metals accelerates cracking.⁽¹⁹⁾ The action of the addition of noble metals can thus be explained as one of local stimulation of anodic attack, which gives rise to rapid crack formation.

Nickel does not crack in boiling magnesium chloride, even when stressed above its yield point. It has also been shown that cracks do not propagate in nickel.⁽²⁰⁾ This work has now been extended, and results with bonded samples show that some 20% nickel stainless steels, as well as nickel-base alloy No. 1, are resistant to already nucleated cracks. The fraction of the total cracks, formed in the susceptible part of the compound specimens, which fail to penetrate into the steels increases with increasing time to failure when they are tested alone. Thus, it seems reasonable to assume that the longer times to failure are caused by increasing difficulty in finding a crack-susceptible site during initial corrosion which precedes the propagation period. It is possible that crack initiation, as such, does not exist as a separate step during stress corrosion cracking in austenitic stainless steels. Assuming that the time lapse before the onset of propagation is actually spent in exposing a susceptible path to the corrosive medium, then all that is needed would be sufficient general corrosion or film destruction, and no true nucleation

is required. It is also possible that there is no great difference between the initial corrosion or film destruction at the surface of a crack-sensitive or a crack-resistant material. It remains to be established whether the "existing paths" of rapid localized corrosion are real, and, if so, how they are formed in a stressed metal.

In previous discussions it was concluded that cracks propagate gradually and considerable evidence was given in favor of an electrochemical mechanism.⁽¹³⁾ The present work is not in disagreement with such a mechanism, but the results suggest that the way in which corrosion can proceed at sufficiently high current densities to account for observed rates of cracking is not mainly one of strain-induced depolarization at the crack roots. If this were so, then it would be expected that the results in Table XI would firstly show strain-induced depolarization at the higher current densities, and, secondly, that there would be a correlation between crack susceptibility and the depolarization observed. These requirements are not fulfilled. An alternative explanation for selective corrosion may be the formation of composition gradients within the stressed alloy. Such segregation of solute atoms has recently been reported by Swann and Nutting.^(21,22) These authors also demonstrated selective attack in polishing solutions at sites of imperfection and changes in composition, but they interpret the contribution of localized corrosion as only a part of stress corrosion cracking, the other step consisting of physical fracture. In stainless steel it is believed that the mechanism is entirely electrochemical.^(13,14) This does not exclude the possibility of continuous removal of metal to form cracks by selective corrosion according to Swann and

Nutting, it only omits the mechanical step. The enriched alloy zone in the region of stacking faults would have to be sufficiently anodic to the matrix to provide a potential difference large enough to account for the high current densities required for crack propagation. Along the same lines, Uhlig and White⁽⁶⁾ suggested the setting up of strongly cathodic areas by diffusion of nitrogen to lattice imperfection sites during plastic deformation. The influence of structural and composition changes, caused during straining of alloys on stress corrosion will have to be evaluated more fully before a better picture of the mechanism of cracking can be obtained.

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TABLE I
COMPOSITION OF VACUUM-MELTED ALLOYS WITH 20% NICKEL

	Alloy No.	<u>Ni</u>	<u>Cr</u>	<u>C</u>	<u>N</u>	<u>Mo</u>	<u>Mn</u>	<u>O</u>	<u>S</u>	<u>P</u>
1	(VM 201)	19.9	16.2	.013	.010			.0015	.011	.003
2	(VM 203)	19.9	16.1	.0013	.011	1.51		.027	.007	.003
3	(VM 466)	20.1	16.1	.06	.0007		1.47	.0032	.011	.003
4	(VM 205)	20.0	16.0	.012	.016		1.45	.0037	.011	.003
4A	(VM 467)	20.1	15.9	.05	.05		1.7			
5	(VM 266)	20.0	16.2	.104	.0025			.002	.0105	.0005
6	(VM 462)	20.1	15.9	.048	.0003					
7	(VM 463)	19.9	16.0	.046	.0008	1.48				

TABLE II
STRESS-CORROSION AND MECHANICAL PROPERTIES OF VACUUM-MELTED
ALLOYS WITH 20% NICKEL

	Alloy No.	Main Impurity	2% Yield Stress psi	U.T.S., psi	Test Stress psi	No. of Tests	Hours to Failure		
							Min.	Max.	Ave.
1	(VM 201)	-	12,800	63,000	40,000	6	>250	-	>250
2	(VM 203)	1.5 % Mo	18,500	65,000	40,000	2	9	13	11
3	(VM 460)	1.47% Mn	-	-	40,000	2	>200	-	>200
4	(VM 205)	1.45% Mn	16,950	69,000	40,000	2	>130	-	>130
4A	(VM 467)	1.7% Mn, 0.05% N	-	-	40,000	2	125	>130	-
5	(VM 266)	.1 % C	-	-	40,000	2	>145	-	>145
6	(VM 462)	-	-	-	40,000	2	>350	-	>350
7	(VM 463)	1.48% Mo	-	-	40,000	3	10	17	13

TABLE III
COMPOSITION OF AIR-MELTED ALLOYS WITH 20% NICKEL

Alloy No.	NI	Cr	C	N	Mo	Mn	Si	Al	O	S	P
8 (AM 7293)	20.0	16.6	.013	.021	1.03	1.44	.49	.042	.012	.007	.002
9 (AM 7294)	19.9	16.7	.013	.026	2.01	1.40	.49	.040	.011	.007	.002
10 (AM 7295)	20.0	16.6	.024	.019	2.97	1.48	.47	.054	.014	.008	.002
11 (AM 7296)	20.0	16.4	.015	.015	4.00	2.00	.47	.049	.014	.008	.002
12 (AM 7297)	19.9	16.5	.013	.018	5.06	1.44	.45	.059	.018	.009	.002
13 (AM 7298)	20.0	16.5	.011	.023	2.01	3.77	.45	.069	.013	.011	.002
14 (AM 7299)	20.0	16.4	.015	.020	4.07	3.99	.47	.074	.0077	.010	.002
15 (AM 7300)	19.8	16.7	.017	.028	2.02	6.49	.46	.084	.0032	.010	.002
16 (AM 7301)	19.9	16.6	.014	.033	3.97	6.29	.46	.084	.0081	.009	.002
17 (AM 7329)	19.6	16.7			.002	1.56	.47				

TABLE IV

STRESS CORROSION PROPERTIES OF AIR-MELTED ALLOYSWITH 20% NICKEL

Alloy No.	Test Stress, psi.	No. of Tests	Hours to Failure		
			Min.	Max.	Ave.
8 (AM 7293)	40,000	2	7	9	8
9 (AM 7294)	40,000	2	8	12	10
10 (AM 7295)	40,000	2	8	12	10
11 (AM 7296)	40,000	2	10	14	12
12 (AM 7297)	40,000	3	10	20	14
13 (AM 7298)	40,000	3	4	9	7
14 (AM 7299)	40,000	2	7	8	8
15 (AM 7300)	40,000	2	3	5	4
16 (AM 7301)	40,000	2	5	7	6
17 (AM 7329)	40,000	2	>170	--	>170

TABLE V

COMPOSITION OF LEVITATION-MELTED ALLOYS WITH 20% NICKEL

Alloy No.	Alloy Used as Starting Material	% C Total	% Mo Total	% Mn Total	% N Total
18 (G 182)	2. 16% Cr 20% Ni 1.5% Mo	.2	1.51	--	.011
19 (G 183)	2. 16% Cr 20% Ni 1.5% Mo	.36	1.51	--	.011
20A (G 199)	1. 16% Cr 20% Ni .013% C	.013	.92	--	--
20B (G 203)	1. 16% Cr 20% Ni .013% C	.013	.62	--	--
20C (G 201)	1. 16% Cr 20% Ni .013% C	.013	.17	--	--
20D (G 200)	1. 16% Cr 20% Ni .013% C	.013	.074	--	--
20E (G 202)	1. 16% Cr 20% Ni .013% C	.013	.028	--	--
21 (G 170)	5. 16% Cr 20% Ni .1% C	.104	1.4	--	.0025
22 (G 171)	5. 16% Cr 20% Ni .1% C	.104	--	1.45	.0025
23 (G 172)	5. 16% Cr 20% Ni .1% C	.104	--	1.35	.12
24 (G 173)	5. 16% Cr 20% Ni .1% C	.104	--	1.45	.03
25 (G 174)	4. 16% Cr 20% Ni 1.45% Mn	.012	--	1.45	.12
26 (G 175)	4. 16% Cr 20% Ni 1.45% Mn	.012	--	1.45	.032
27 (G 176)	4. 16% Cr 20% Ni 1.45% Mn	.25	--	1.45	.016
28 (G 177)	4. 16% Cr 20% Ni 1.45% Mn	.34	--	1.45	.016
29 (G 178)	5. 16% Cr 20% Ni .1% C	.1	--	--	.0025
30 (G 179)	5. 16% Cr 20% Ni .1% C	.2	--	--	.0025

TABLE VI

STRESS CORROSION AND MECHANICAL PROPERTIES OF LEVITATION-MELTED
ALLOYS WITH 20% NICKEL

Alloy No.	Main Impurity	2% Yield Stress psi	U.T.S., psi	Test Stress psi	No. of Tests	Hours to Failure		
						Min.	Max.	Ave.
18	1.51% Mo, .2% C	29,300	86,500	40,000	2	14	21	18
19	1.51% Mo, .36% C	41,500	105,700	40,000	2	>200	-	>200
19	-	-	-	50,000	2	80	106	93
20A	.92% Mo	-	-	40,000	2	28	32	30
20B	.62% Mo	-	-	40,000	2	15	34	25
20C	.17% Mo	-	-	40,000	2	40	51	45
20D	.074% Mo	-	-	40,000	2	85	121	103
20E	.028% Mo	-	-	40,000	2	66	90	78
21	1.4% Mo, .1% C	25,900	80,100	40,000	2	9	13	11
22	1.45% Mn, .1% C	25,000	78,100	40,000	2	>150	-	>150
23	1.35% Mn, .1% C, .12% N	38,300	95,800	40,000	3	103	149	120
24	1.45% Mn, .1% C, .03% N	25,500	85,900	40,000	3	140	>200	>200
25	1.45% Mn, .01% C, .12% N	37,800	87,200	40,000	2	30	30	30
26	1.45% Mn, .01% C, .03% N	21,200	66,200	40,000	2	54	87	70
27	1.45% Mn, .25% C, .016% N	-	-	40,000	2	>150	-	>150
28	1.45% Mn, .34% C, .016% N	-	-	40,000	2	>150	-	>150
29	.1% C, .0025% N	-	-	40,000	2	>150	-	>150
30	.2% C, .0025% N	-	-	40,000	2	>150	-	>150

TABLE VII

COMPOSITION OF LEVITATION-MELTED ALLOYS WITH 14% NICKEL

Starting Material: 31(VM 379), 14.3% Ni, 16.3% Cr, .0013% C,
 .002% N, <.003% Ti, .002% Mn,
 .03% Si, .003% Al, .009% O,
 .007% S, .0023% P

Alloy No.	% C	% N	% Ti	% Mn	% Si	% Al	% Pt	% Cb	% Sn
32 (G 1)			.05						
33 (G 2)			.18						
34 (G 3)			.61						
35 (G 4)				.45					
36 (G 5)					.14				
37 (G 6)						.042			
38 (G 7)		.046				.014			
39 (G 8)		.007							
40 (G 9)		.12							
41 (G 10)		.041	.05						
42 (G 11)		.049	.18						
43 (G 12)		.047	.31						
44 (G 13)		.036	.65						
45 (G 14)		.041		1.84					
46 (G 15)		.041				.094			
47 (G 16)		.044			.45				
48 (G 102)					.53				
49 (G 103)			.37						
50 (G 104)				1.96					

(Continued)

[illegible]

TABLE VIII

STRESS CORROSION AND MECHANICAL PROPERTIES OF ALLOYS WITH 14% NICKEL

Alloy No.	Addition	.2% Yield Stress psi.	U.T.S., psi.	Test Stress psi.	No. of Tests	Hours to Failure		
						Min.	Max.	Ave.
31 (VM 379)	--	17,400	75,700	35,000	2	42	78	60
31	--	--	--	40,000	2	28	40	34
31	--	--	--	48,000	2	26	30	28
32	.05% Ti	--	--	48,000	3	25	90	50
33	.18% Ti	--	--	48,000	5	11	22	17
34	.61% Ti	--	--	48,000	5	12	34	23
35	.45% Mn	--	--	40,000	2	13	17	15
35	.45% Mn	--	--	48,000	3	12	15	14
36	.14% Si	--	--	48,000	4	21	37	29
37	.042% Al	--	--	48,000	4	14	24	18
38	.046% N	--	--	48,000	4	10	17	13
39	.007% N	--	--	40,000	3	20	56	40
39	.007% N	--	--	35,000	3	40	74	54
40	.12% N	--	--	48,000	3	11	16	14
41	.05% Ti, .04% N	--	--	48,000	2	9	14	12
41	.05% Ti, .04% N	--	--	40,000	1	--	--	29
42	.18% Ti, .049% N	--	--	48,000	2	16	32	24
42	.18% Ti, .049% N	--	--	40,000	1	--	--	25
43	.31% Ti, .047% N	--	--	48,000	4	6	30	17
43	.31% Ti, .047% N	--	--	40,000	1	--	--	41
44	.65% Ti, .036% N	--	--	48,000	3	16	25	21
45	1.84% Mn, .041% N	--	--	48,000	5	35	59	48
46	.09% Al, .041% N	--	--	48,000	4	14	34	22

TABLE VIII (Continued)

Alloy No.	Addition	.2% Yield Stress psi.	U.T.S., psi.	Test Stress psi.	No. of Tests	Hours to Failure		
						Min.	Max.	Ave.
47	.45% Si, .044% N	--	--	48,000	3	20	42	31
48	.53% Si	--	--	33,400	2	47	79	63
49	.37% Ti	--	--	33,400	2	40	50	45
50	1.96% Mn	--	--	33,400	1	--	--	43
51	.012% N	--	--	33,400	2	43	72	57
52	.03% N	--	--	33,400	2	30	70	50
53	.48% Si	--	--	48,000	3	27	41	35
54	1.93% Mn	--	--	48,000	3	16	45	23
55	.063% Al	--	--	48,000	3	17	24	21
56	.023% C	16,700	76,600	48,000	2	32	41	37
57	.44% Ti	--	--	48,000	3	18	41	27
58	.096% C	24,700	89,400	40,000	2	>160	--	>160
58	--	--	--	48,000	2	>160	--	>160
58	--	--	--	60,000	2	>160	--	>160
59	.20% C	43,700	88,600	40,000	2	>200	--	>200
59	.20% C	43,700	88,600	50,000	2	>250	--	>250
59	.20% C	43,700	88,600	59,000	2	>110	--	>110
60	.40% C	60,900	126,800	40,000	2	>220	--	>220
60	.40% C	60,900	126,800	59,000	2	>190	--	>190
60A	.1% C (also see analyses)	--	--	60,000	2	>160	--	>160
60B	.1% C	--	--	60,000	2	>160	--	>160
60C	.1% C	--	--	60,000	2	>160	--	>160
60D	.1% C	--	--	60,000	2	>160	--	>160
60E	.1% C	--	--	60,000	2	>160	--	>160
61	.1% Pt	--	--	--	2	4	4	4

TABLE VIII (Continued)

Alloy No.	Addition	2% Yield Stress psi.	U.T.S., psi.	Test Stress psi.	No. of Tests	Hours to Failure		
						Min.	Max.	Ave.
62	.1% Pt	--	--	--	2	3	4	4
63	.0015% Sn	--	--	50,000	2	66	78	72
64	.17% Sn	--	--	50,000	2	68	136	102
65	.33% Sn	--	--	50,000	2	25	27	26
66	.44% Sn	--	--	50,000	2	16	40	28

TABLE IX
COMPOSITION OF HIGH NICKEL ALLOYS

<u>Alloy No.</u>	<u>Type</u>	<u>% Ni</u>	<u>% Cu</u>	<u>% Fe</u>	<u>% Si</u>	<u>% Mn</u>	<u>% C</u>	<u>% Cr</u>	<u>% S</u>	<u>% Al</u>	<u>% Ti</u>	<u>% Cb+Ta</u>
67	"61" Nickel	95.80	.02	.19	.47	.27	.07	-	.005	-	2.69	-
68	Nickel-Copper Alloy	65.97	30.67	.07	.52	.35	.01	-	.005	-	2.33	-
69	Nickel-base Alloy No. 1	73.86	.04	7.09	.12	.16	.02	16.35	.005	-	-	2.33
70	Nickel-base Alloy No. 2	73.16	.05	7.05	.30	.61	.05	15.09	.007	.69	2.26	.71
71*	"A" Nickel											

* Not analyzed.

TABLE X

STRESS CORROSION PROPERTIES OF HIGH NICKEL ALLOYS

Alloy No.	Test Stress, psi.	Hours to Failure
67	35,000	>500
68	35,000	>500
69	35,000	>500
70	35,000	>500
71	35,000	>500

TABLE XI

EFFECT OF STRAIN ON ANODIC POLARIZATION
AT LIQUID FLOW RATES OF 50 AND 100 cm/sec

Material	C. D. 2 amps/cm	Cathode Used as in Fig. 1	Strain Rate %/sec	Point of Potential Measurement, as in Fig. 1	Depolarization Upon Straining, mv.
Commercial 304: 18Cr-8Ni	.35 to .7	A + B	7, 12 >100	1 2	0 0
	.35	A	12	1 2	0 75
	.03	A + B	12	1 2	50 50
Nickel "61"	.35 to .7 .03	A + B	12	1 or 2	0 5, 10, 20
Nickel "A"	.35 to .7 .03	A + B	12	1 or 2	0 85, 75
Nickel-base Alloy No. 1	.35 to .7 .03	A + B	12	1 or 2	0 30
VM 462 "Pure" 16Cr-20Ni	.35 to .7 .03	A + B		1 or 2	0 20, 35, 50

TABLE XI

(Continued)

Material	C. D. ² amp./cm	Cathode Used as in Fig. 1	Strain Rate %/sec	Point of Potential Measurement, as in Fig. 1	Depolarization Upon Straining, mv.
VM 463 "Pure" 16 Cr-20 Ni-1.5Mo	.35 to .7 .03	A + B	12	1 or 2	0 80
VM 466 "Pure" 16 Cr-20Ni-1.5Mn	.35 to .7 .03	A + B	12	1 or 2	0 45,55

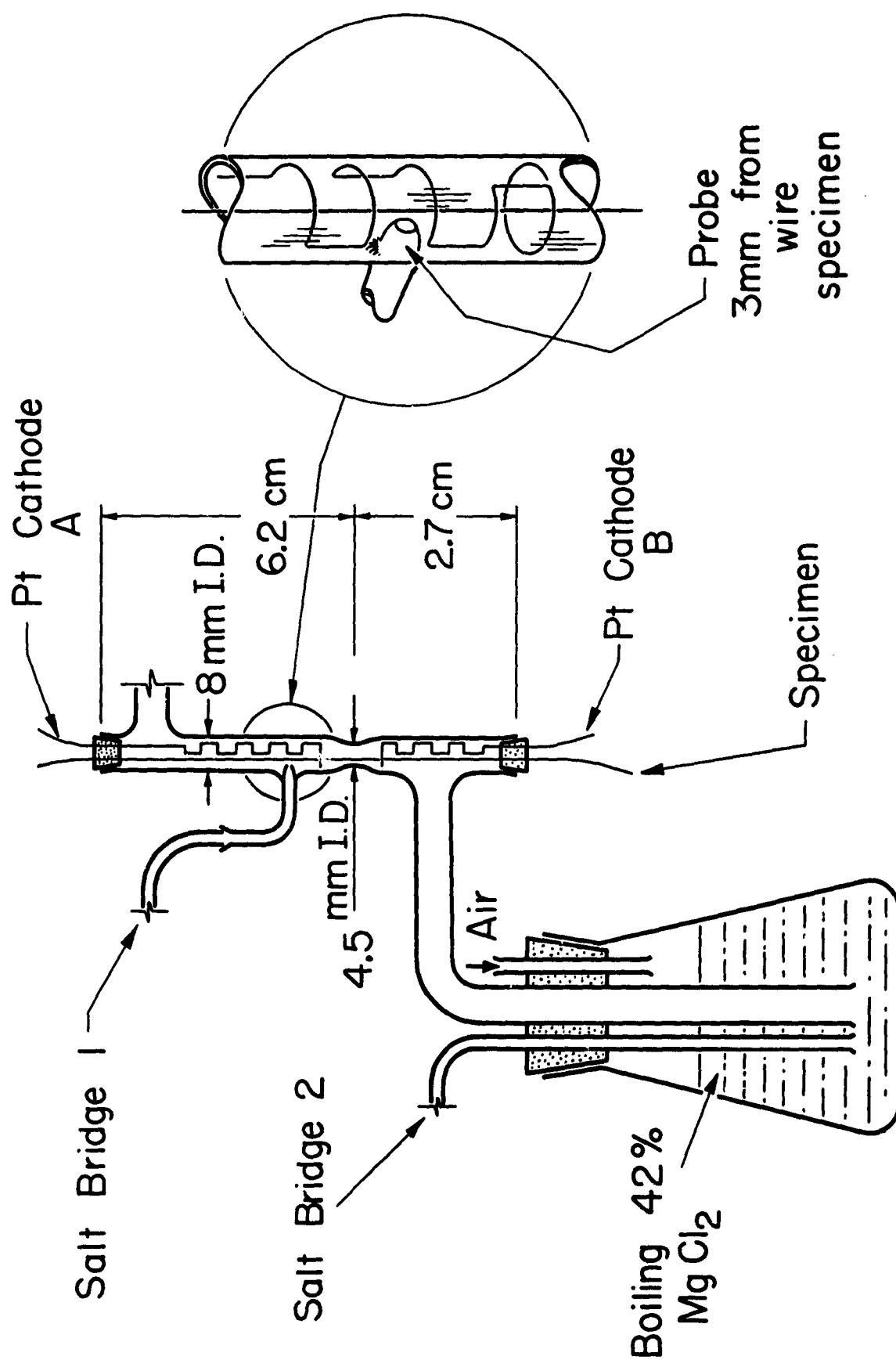


Fig. 1. Apparatus for Polarization Measurements.

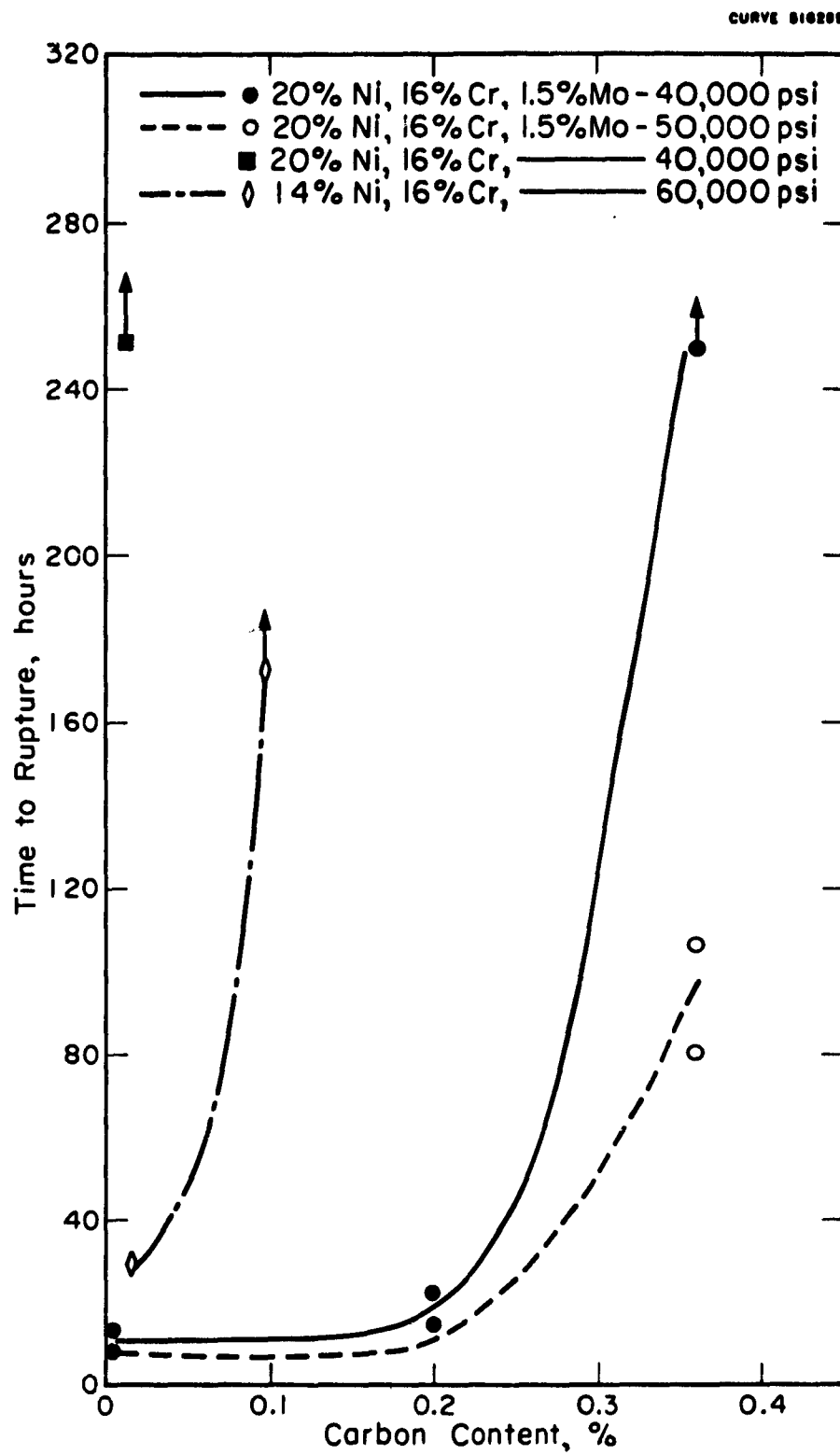


Fig. 2. Effect of Carbon on Stress Corrosion Cracking.

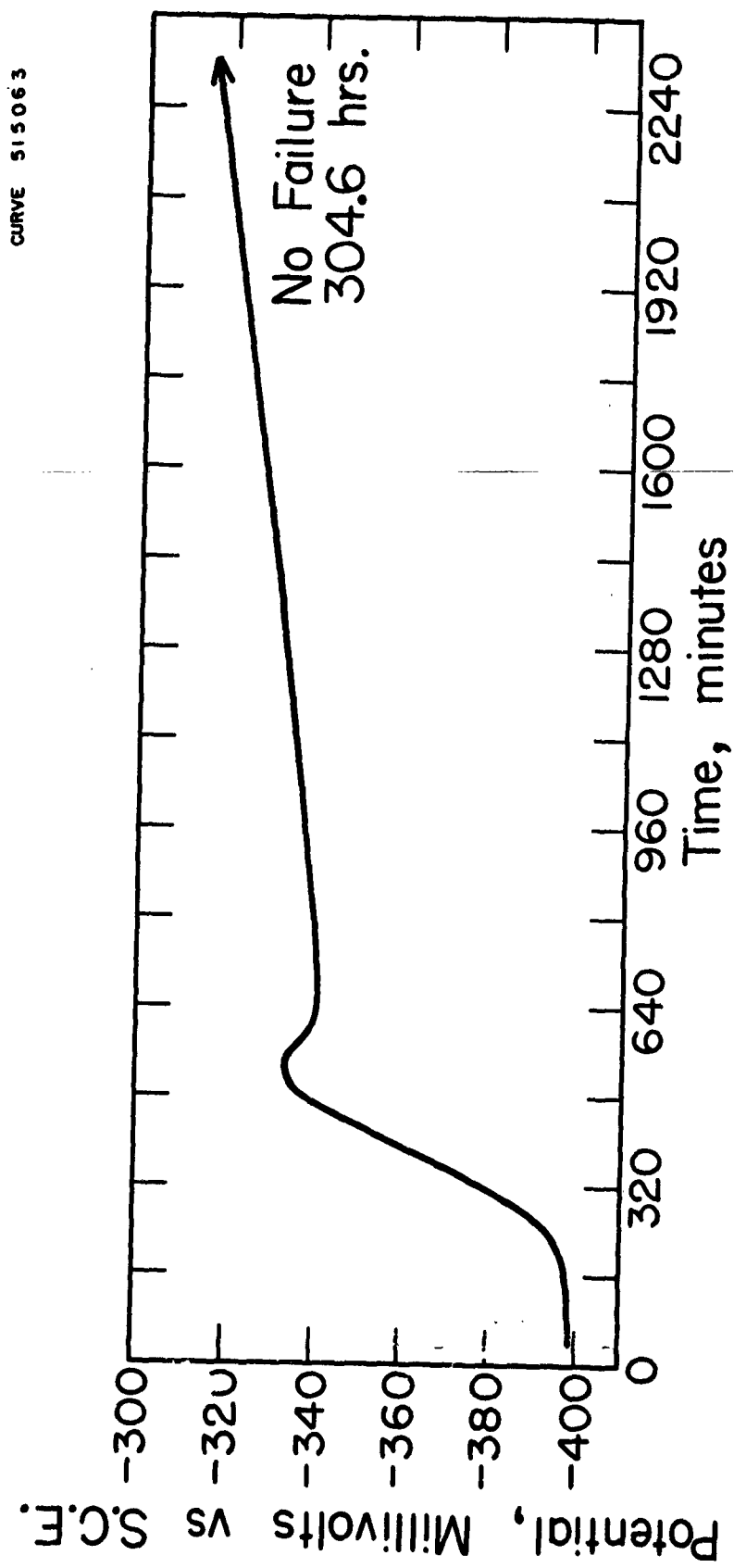


Fig. 3. Potential-time behavior of "pure" alloy with 20% Ni.

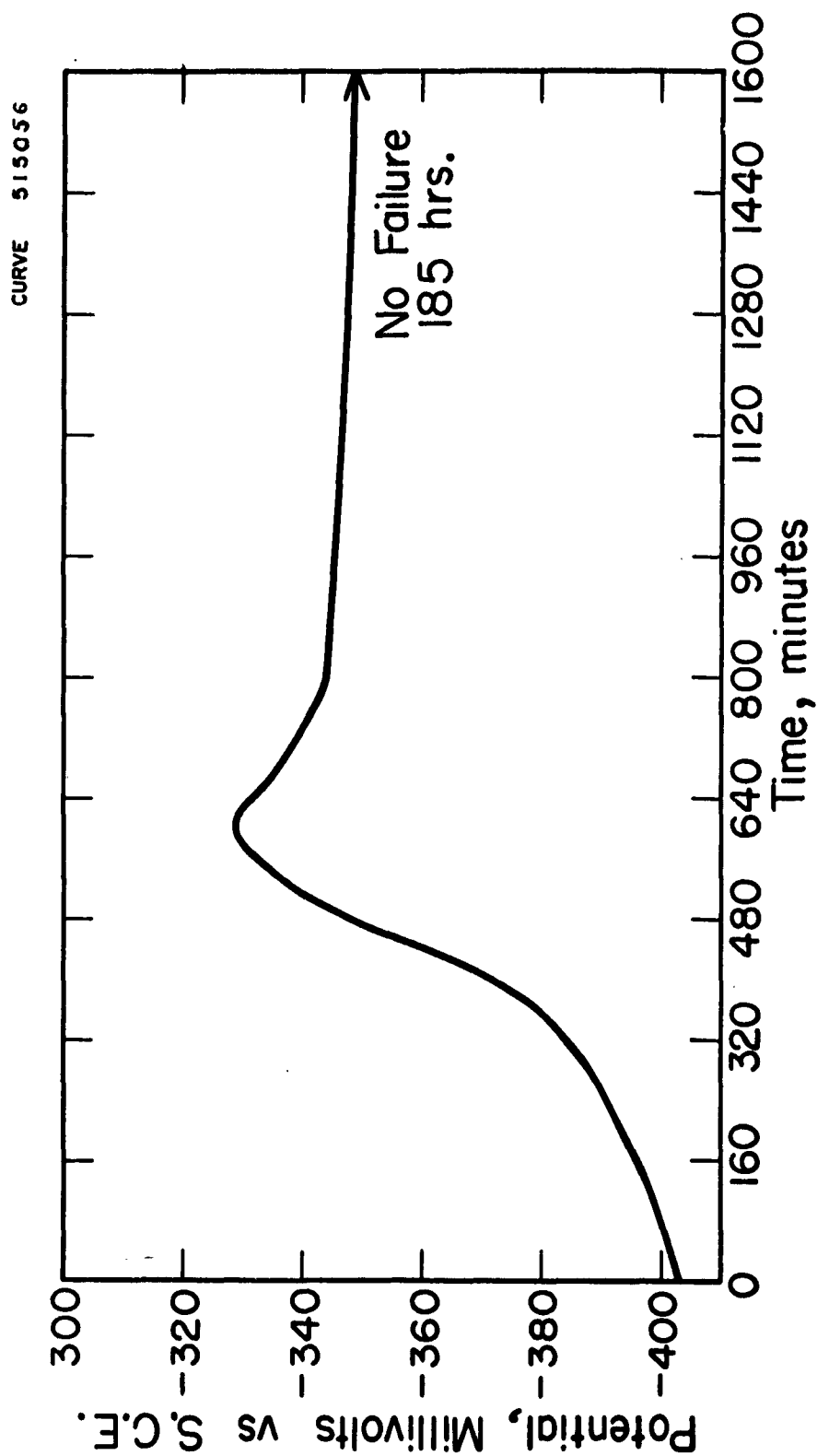


Fig. 4. Potential-time behavior of 20% Ni alloy with Mn.

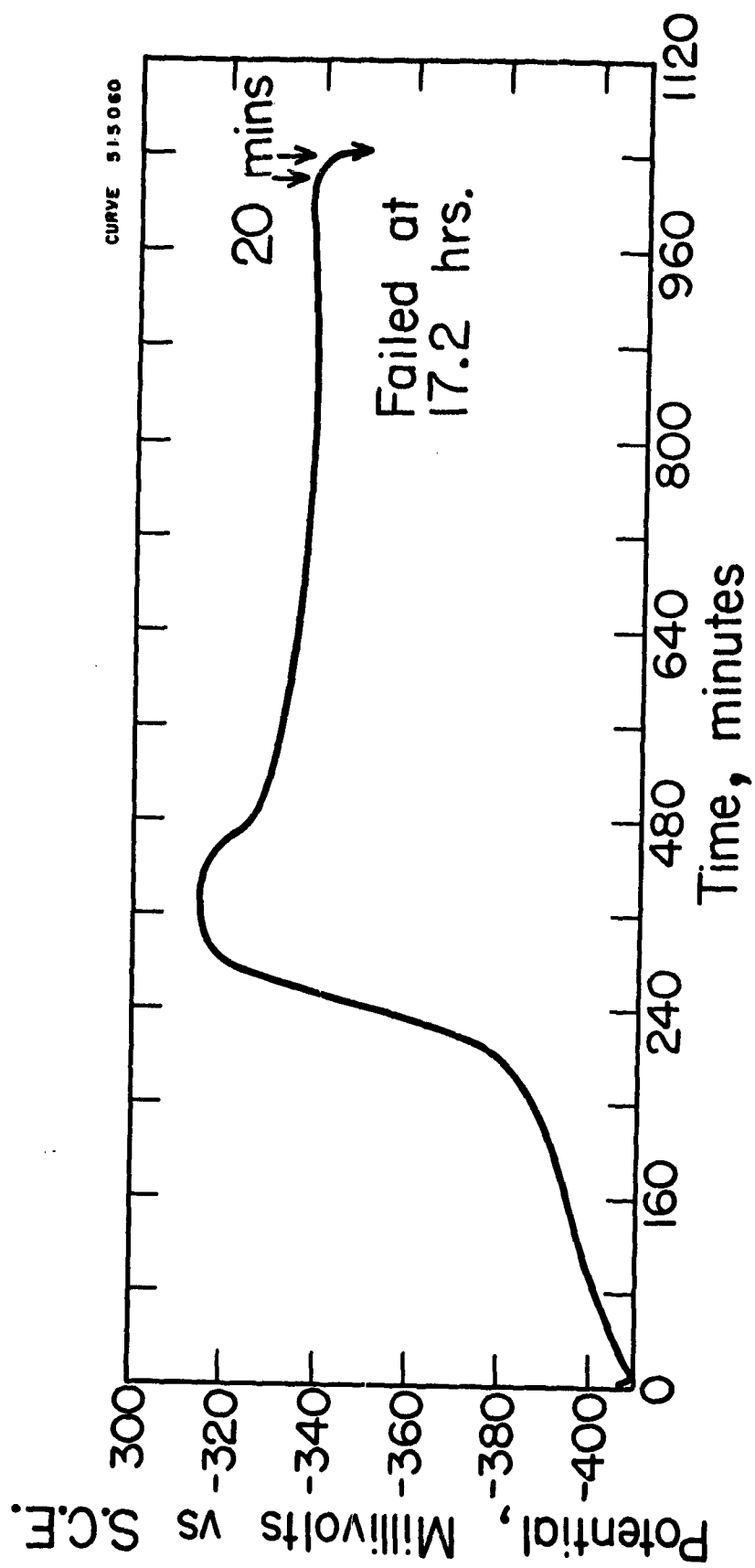


Fig. 5. Potential-time behavior of 20% Ni alloy with N.

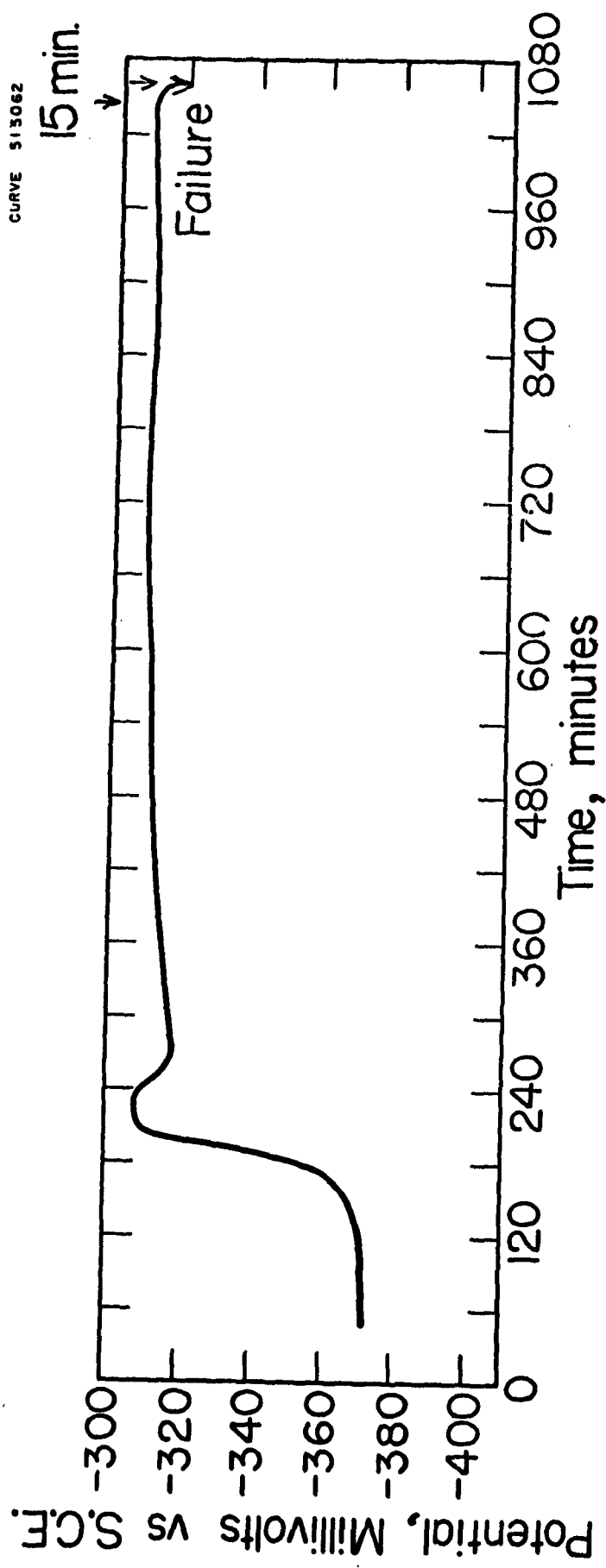


Fig. 6. Potential-time behavior of 20% Ni alloy with Mo.

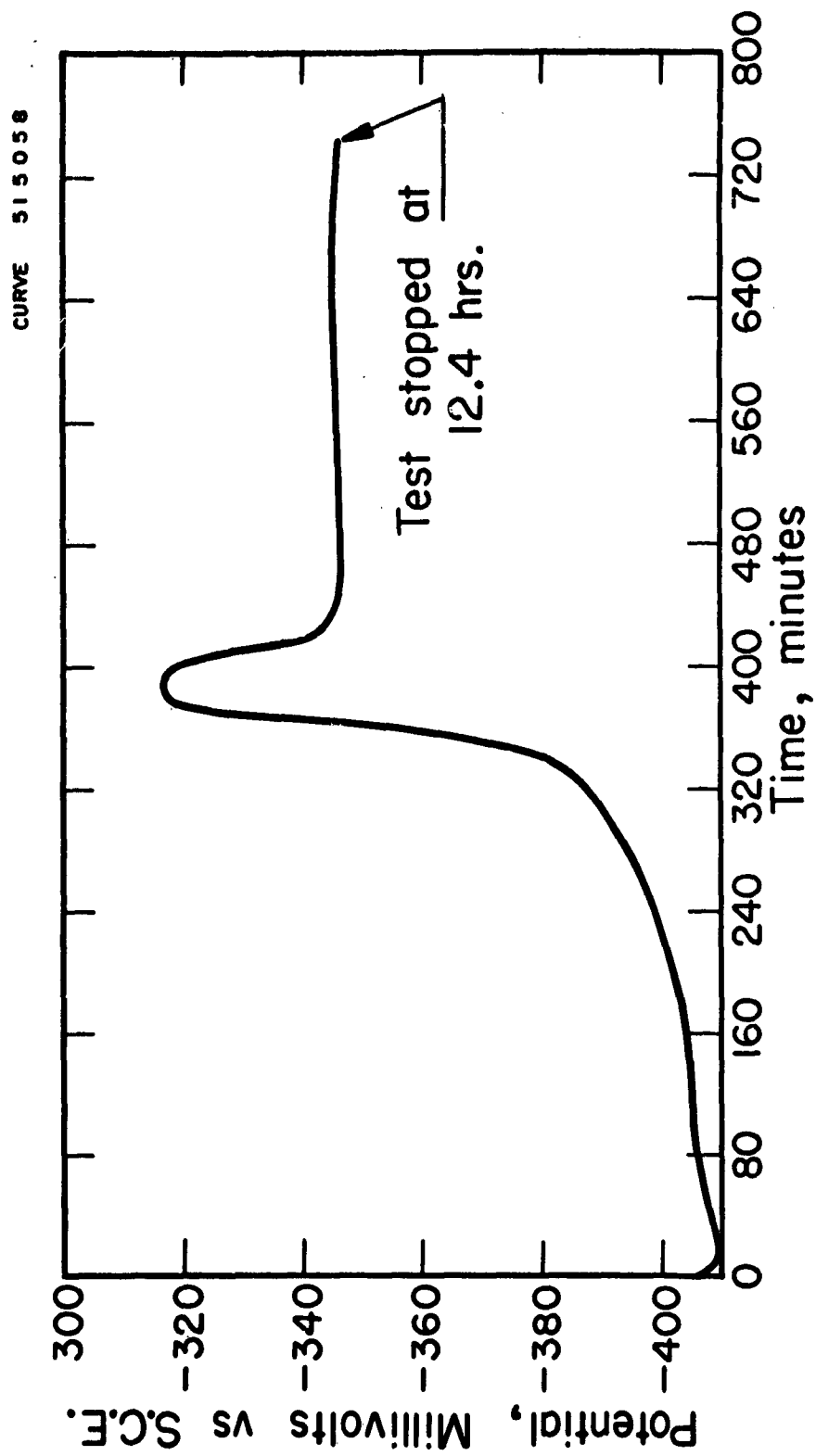


Fig. 7. Potential-time behavior of 20% Ni alloy with N.

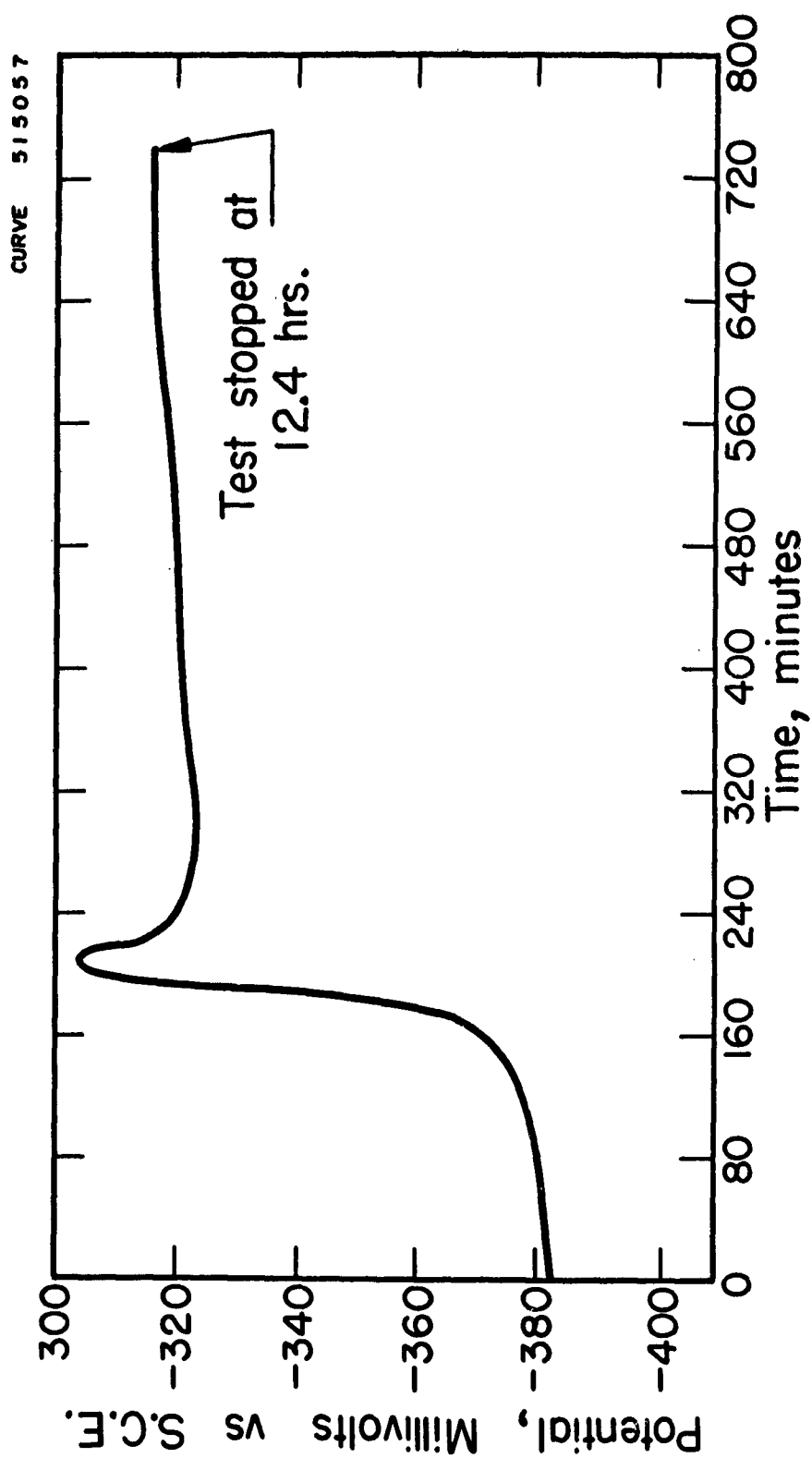


Fig. 8. Potential-time behavior of 20% Ni alloy with Mo.

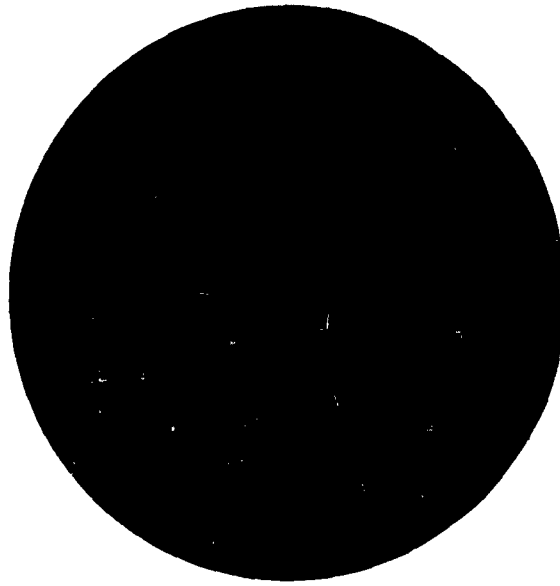


Fig. 9. Method of Mounting Wire Specimens.

X3

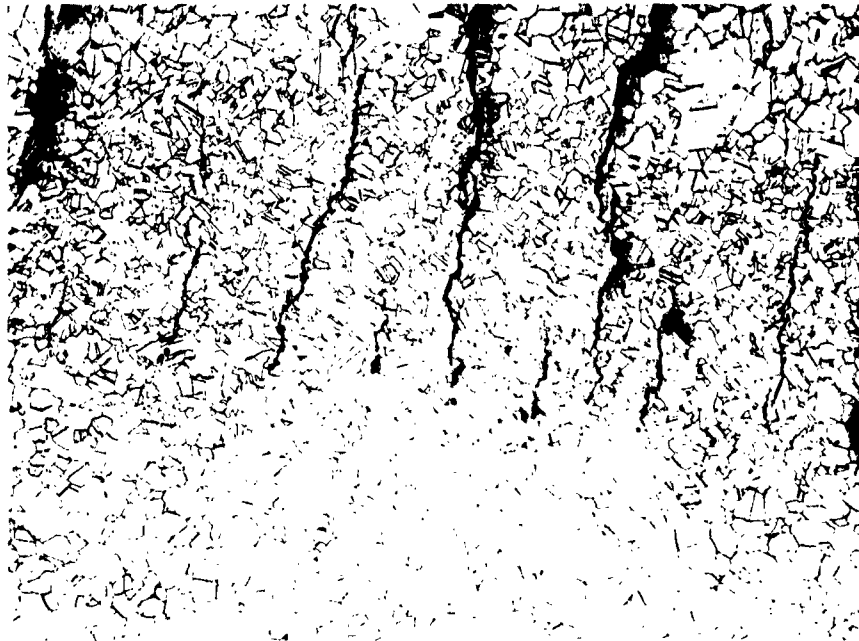


Fig. 10. Cracks Stopping upon Reaching "Pure"
16 Cr-20 Ni Alloy.

X100

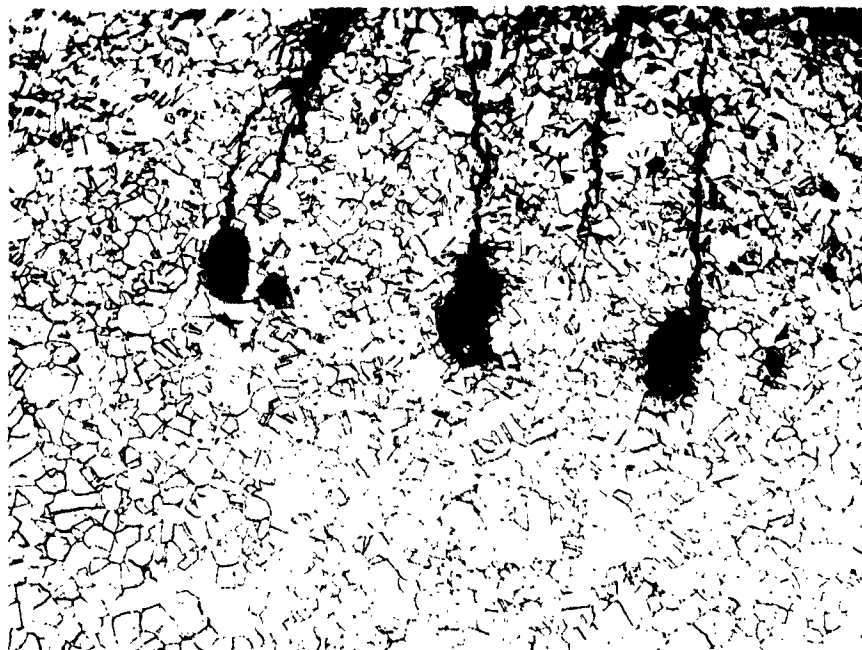


Fig. 11. Pitting at Roots of Cracks in "Pure"
16 Cr-20 Ni Alloy.

X100

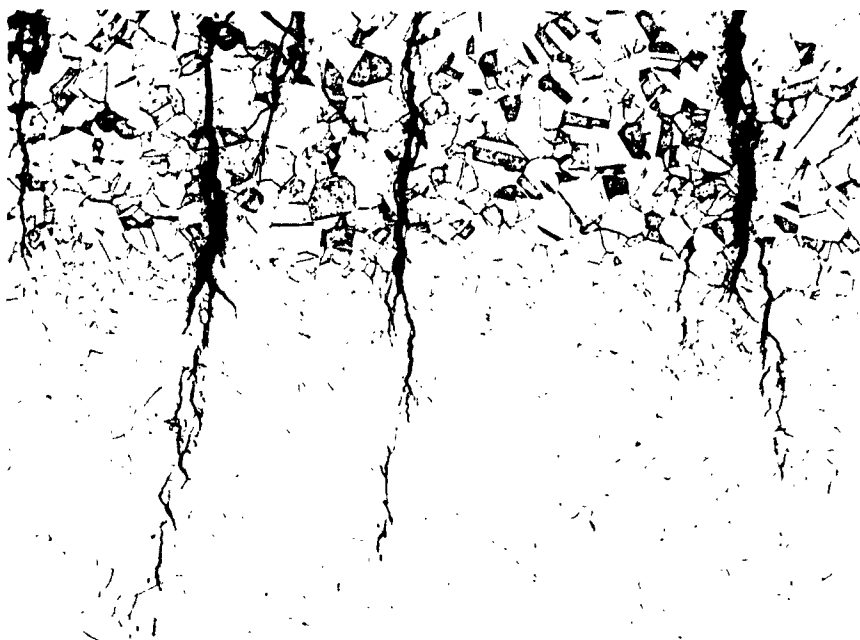


Fig. 12. Cracks Penetrating into 20 Ni Alloy with
1.5 Mo.

X100

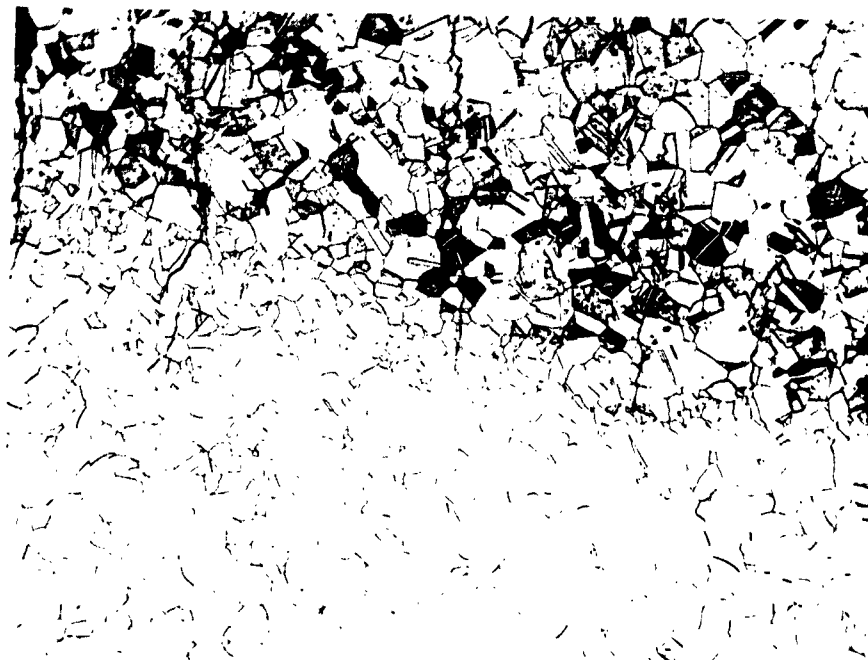


Fig. 13. Cracks Stopping upon Reaching 20 Ni Alloy
with 1.5 Mo. X100