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**LITERATURE SURVEY ON WELD-METAL CRACKING**

The first of a series of reports  
on  
**WELD-METAL CRACKING**

A. J. WILLIAMS  
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BATTELLE MEMORIAL INSTITUTE

AUGUST 1952

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WRIGHT AIR DEVELOPMENT CENTER

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*August 1952*

*Materials Laboratory*  
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Wright Air Development Center  
Air Research and Development Command  
United States Air Force  
Wright-Patterson Air Force Base, Ohio

## FOREWORD

This report was prepared by the Battelle Memorial Institute under U. S. Air Force Contract No. AF 33(038)-12619. The contract was initiated under the research and development project identified by Research and Development Order No. R607-43, "Fusion Welding Processes and Equipment", and was administered under the direction of the Materials Laboratory, Research Division, Wright Air Development Center, with Major Robert E. Bowman acting as project engineer. This report is the first of a series to be issued on this project. Others in the series will be issued as research progresses.



## ABSTRACT

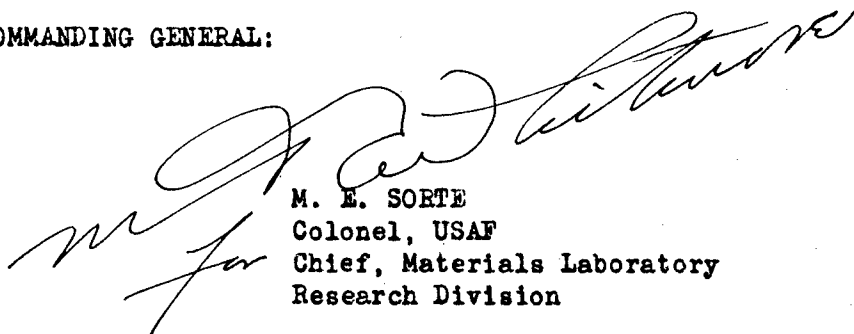
As the initial part of an investigation for Wright Air Development Center on the causes of cracking in high-strength weld metal, a literature survey was made. Work on wrought and cast steel was included in this survey, since it is the most prolific source of information on this subject.

Of the various phenomena which might be pertinent, hot tearing, hot shortness, blue brittleness, and hydrogen embrittlement appeared to be closely related to the causes of weld-metal cracking. Investigators agree that most weld-metal cracking initiates at high temperatures in the boundaries between grains or dendrites. This type of cracking is generally attributed to the presence on these boundaries of low-melting-point or "brittle" compounds — complex sulfides, for example. Most types of low-temperature cracks in weld metal appeared to be associated with rapid cooling rates and hydrogen. As a result of this survey, the experimental program on this investigation will involve weld-metal cracking tests, studies of grain-boundary constituents which might cause cracking, preparation of special weld-metal compositions, and tests to determine the temperature at which cracking occurs.

## PUBLICATION REVIEW

Manuscript Copy of this report has been reviewed and found satisfactory for publication.

FOR THE COMMANDING GENERAL:



M. E. SORTE  
Colonel, USAF  
Chief, Materials Laboratory  
Research Division

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# LITERATURE SURVEY ON WELD-METAL CRACKING

## INTRODUCTION

A literature survey was set up as the initial part of an investigation on the causes of cracking in high-strength weld metal. The purpose of the survey was to determine the present status of knowledge on this subject as a guide to the experimental work.

Weld-metal cracking has long been a problem for the welding engineer and is particularly acute with the use of high-strength steels in aircraft manufacture. In this industry, the stringent service requirements have made it essential that weldments be free of defects to a greater degree than is necessary for many comparatively low-strength welded fabrications.

In a limited number of applications, cracking in high-strength steel welds has been reduced by using special procedures or with especially favorable designs. However, such measures may not be practical in the welding of large steel assemblies such as aircraft landing gears. Even the use of large safety factors may result in oversize weld deposits which, in some cases, increase the possibility of weld-metal cracking. This situation has shown the need for more fundamental information on the mechanisms which cause weld-metal cracking in high-strength steels. [This information is needed on welds in aircraft steels and other high-strength ferritic steels.]

In recognition of this situation, Wright Air Development Center authorized this investigation on the fundamental causes of, and the methods of controlling, weld-metal cracking.

The literature survey covered about 1450 published papers on weld-metal cracking and related subjects, of which the most pertinent are listed in the bibliography in the appendix of this report. The survey also included unpublished information obtained through visits and correspondence with various authorities on welding problems.

The extensive studies made of austenitic weld deposits in recent years have led to a possible means of preventing cracking in these materials. Many users of stainless steel electrodes found that weld-metal cracking could be minimized by introducing a small amount of delta ferrite into the microstructure by properly adjusting the composition. The situation with weld metal deposited with high-strength ferritic electrodes is quite different. Such metal may be delta ferritic or austenitic during and directly after deposition, depending on the composition, and any relation here between structure and cracking appears to have escaped attention. In any event, the literature on high-strength ferritic weld deposits is rather limited, and at the present time it is not known whether cracking can be reduced by adjusting the composition, except possibly by lowering the carbon content and by keeping the sulfur content below a certain maximum.

Considerable disagreement was found in the literature as to what causes weld-metal cracking. Many investigators feel that it is related to hot tearing in castings. Others suggest that the phenomena of hot shortness, phase transformation, and blue brittleness are just as important, though the pertinent question of the temperature at which cracking occurs is without a convincing answer. However, there was general agreement that cracking in weld metal was largely interdendritic or intergranular regardless of the temperature at which it occurred. Very few examples of transgranular cracking were found. Even low-temperature or cold cracks in weld metal were found to be intergranular. Several investigators showed that these cracks were caused by hydrogen associated with rapid cooling rate and retained austenite.

Since most weld-metal cracking is interdendritic or intergranular, it is reasonable to assume that differences in the properties of grain-boundary materials might cause one weld metal to be more crack sensitive than another. In this light, it would be in order to identify grain-boundary segregates in order to correlate their composition with weld-cracking tendency. It is suggested later that these studies could possibly be made using electron-diffraction techniques, the electron microscope, microspectrographic analysis, etc. Some work is now in progress by other investigators in which fundamental causes of cracking in austenitic welds are being studied by using electron-diffraction techniques.

The first part of this report is devoted to a literature review on various phenomena, such as hot shortness, hot tearing, hydrogen embrittlement, etc., which are fundamental factors affecting the ductility of steel. This is presented first as a background for discussing the problem of weld-metal cracking. The factors which cause the cracking could then be better understood. After the section on fundamental factors, various types of cracks encountered in weld metal are described, and a discussion is presented of the known mechanical and metallurgical factors which cause these cracks. The methods used by many investigators to study and remedy cracking are then reviewed.

The literature survey showed that there is considerable information that is not known about the mechanism of weld-metal cracking. Future experimental work on this investigation will be directed toward studies to determine these unknown factors. These will involve weld-metal cracking tests, studies of grain-boundary constituents which might cause the cracking, preparation of special weld-metal composition, and tests to determine the temperature at which cracking occurs.

### FACTORS AFFECTING THE FRACTURE STRENGTH OF NONWELDED STEEL

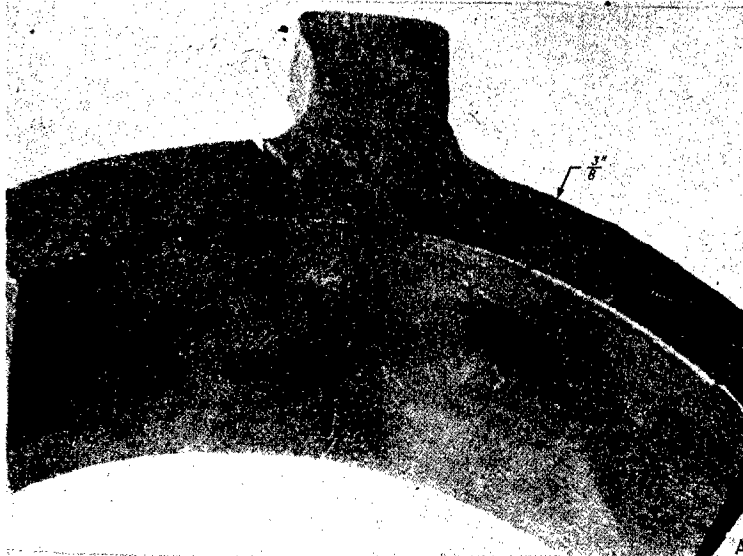
There is considerably more information available on the causes of low strength and embrittlement in nonwelded steel than there is on the causes of weld-metal cracking. This is probably because welding, as we know it today, is a relatively new art. Therefore, as a background for a study of weld-metal cracking, the various factors affecting the ductility of steel should first be considered in view of their possible effect on weld-metal cracking.

In this survey, only cracking in weld metal which occurs during welding or a short time after cooling to room temperature is being considered. The so-called "service cracks" are not discussed in this report. Therefore, only those factors which appeared to have some relation to weld-metal cracking were included in the survey. Information on cracking in heat-affected parent-metal zones, such as underbead cracking, has been omitted from this survey, since this was in itself a broad subject and should be treated separately.

#### Hot Tears and Shrinkage Cavities

Hot tears are cracks which form during cooling of steel when subjected to excessive strains in the temperature range from about 2600 F to about 2200 F. A common example is the tears which occur during restrained

cooling of steel castings in sand molds. These cracks generally have a very irregular and jagged appearance, and no evidence of plastic deformation is visible. A photograph of a casting with a hot tear is shown in Figure 1.



a. Small Casting (Actual Size)



b. Large Casting (2/3 Actual Size)

FIGURE 1. EXTERNAL HOT TEARS IN CASTINGS (A4)

The chief cause of hot-tear formation in steel castings is the fact that free contraction of the metal is hindered by the molds and cores used in manufacture, which results in high local stresses(A2, A12). A difference in cross section of the casting also may be a factor if a hot portion is required to shrink against the resistance of a colder portion.

Hot tears are intergranular in nature and, if open to the air, the fracture surfaces are oxidized. From experimental observations, it has been established that hot tears form at temperatures not greatly below the solidifying temperature of steel. Briggs(A4) believed that the temperature range 2350°F to 2300°F constituted the lower range of hot-tear formation in 0.25 to 0.40 per cent carbon steels. Several investigators(A5, A14, A20) have fixed the critical temperature range of hot tearing at 2370°F to 2550°F, as determined by elongation and reduction-of-area measurements of cast steel made at elevated temperatures. Hot tensile tests of cast steel specimens were made by Hall(A14) in rapid succession immediately after pouring to eliminate effects caused by any reheating treatment. He determined ultimate strength and elongation values for various steels on cooling from the pouring temperature. Data from a few of his tests are shown in Table 1 and plotted as curves in Figure 2. The results of these tests indicate that carbon steels are less susceptible to hot tearing than alloy steels. This agrees with actual experience in the foundry, since alloy steels are quite troublesome in casting, due to their great susceptibility to hot tearing. Chromium in percentages of around 1 per cent appears to increase the tendency, while the combination of nickel and chromium has an even stronger effect.

It appears, from Hall's results(A15), that high-temperature ductility is more important in hot tearing than hot strength, since differences in elongation between the plain-carbon and the alloy steels were more significant.

The composition of a steel casting has much to do with its susceptibility to hot tearing. High-carbon steels are more susceptible to hot tearing than low-carbon steels(A4, A11). The reason for this is that the strength and ductility increase more rapidly in low- than in medium-carbon steels, as the temperature drops during cooling(A11). Hall(A15) showed that sulfur markedly decreased the tensile strength and ductility of steel at the temperature of hot-tear formation, believed by him to be 2400°F. Several investigators(A11, A13) have reported that the maximum effect of sulfur on hot tearing differed, depending upon the form of inclusions. Steel containing round or eutectic-type inclusions was more susceptible to hot cracking than steel containing the crystalline type, the latter being produced by adding 2-½ pounds of aluminum per ton of steel(A31). It would appear from this that the addition of aluminum to some weld metals might have some effect on reducing the type of hot cracking that is related to hot tearing. (The primary effect of aluminum may be to deoxidize the steel.)

TABLE 1. RESULTS OF HOT-STRENGTH TESTS ON CAST STEELS MADE AT INTERVALS DURING COOLING FROM POURING TEMPERATURES(A14)

Type	Composition, per cent	Specimen No.	Skin(1) Temperature, F	Ultimate Stress, psi	Elongation, per cent (4" gage length)
Carbon Steel	C - 0.20, Si - 0.3, Mn - 0.6	1	2448	448	0
		2	2437	762	0
		3	2422	807	0
		4	2412	1340	0
		5	2383	1500	1
		6	2340	1900	3
		7	2289	2380	29
Manganese - Molybdenum	C - 0.33, Si - 0.3, Mn - 2.0, Mo - 0.3	1	2444	382	0
		2	2430	650	0
		3	2397	898	0
		4	2347	1770	0.6
		5	2269	2330	2
		7	2145	3320	27
		8	2095	< 3500	< 3.5
		Chromium - Nickel	C - 0.15, Si - 0.5, Mn - 0.6, Ni - 8.0, Cr - 18.0	1	2368
2	2354			808	0
3	2340			1390	0.5
4	2331			1970	1.3
5	2250			2760	4
7	2151			< 3900	< 22

(1) According to Hall, the skin temperature was more important than the core temperature because the first signs of strength would occur at the skin surface of the cast steel specimens.



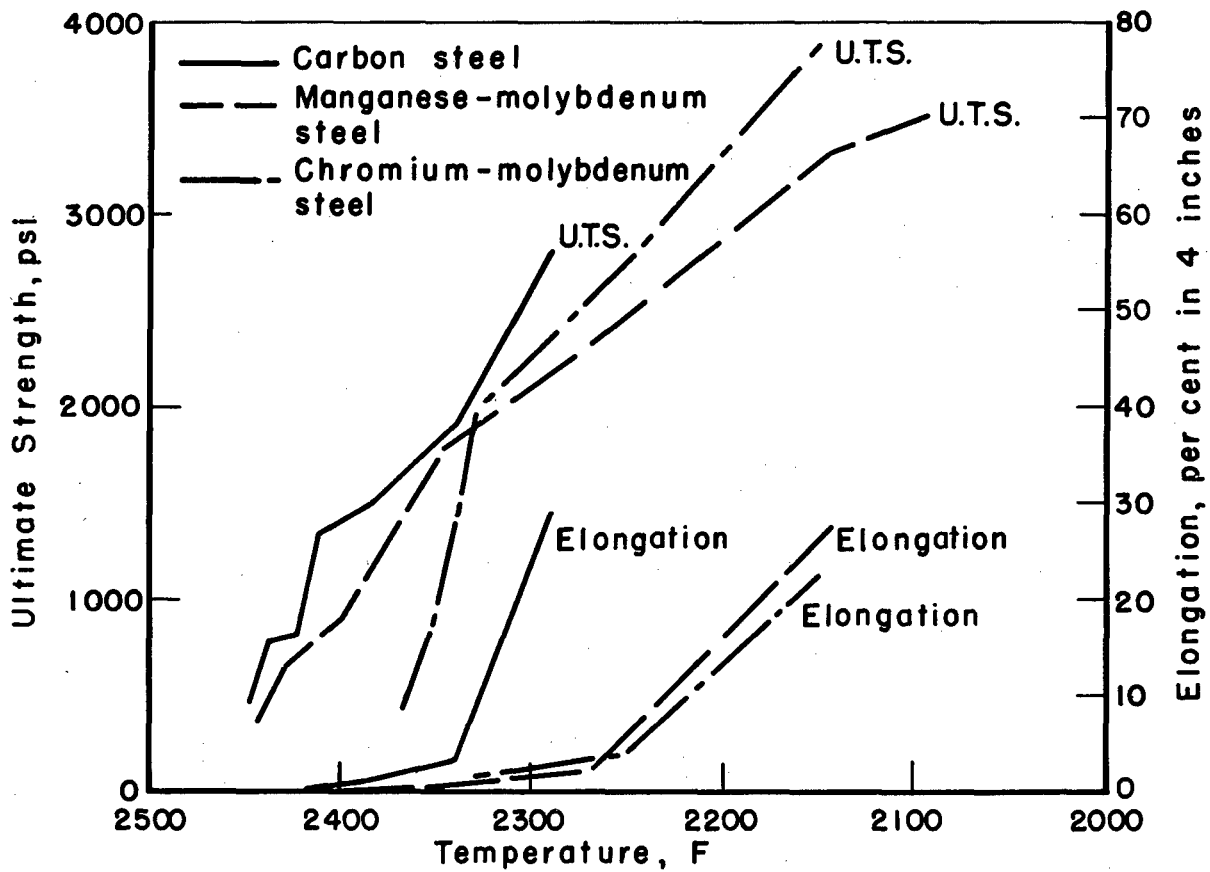


FIGURE 2. HIGH-TEMPERATURE STRENGTH AND DUCTILITY OF CAST STEELS DURING COOLING FROM THE POURING TEMPERATURE (A14)

(The data for these curves are shown in Table I.)

Other factors which appear to affect the hot-tearing tendencies of steel are fluidity, pouring temperature, and grain size. High fluidity makes it possible to fill molds with metal at low pouring temperatures, so that it contains a large amount of solid phase (mushy stage) (A10). The solid phase acts as a "lean" component and thus makes the casting denser and more resistant to hot tearing (A10). No closer definition of what this means was presented. Large grain sizes increase hot-tearing tendencies because of the greater amounts of material in the grain boundaries (A12).

Many investigators have related hot cracking in weld metal to hot tearing in castings. Both seem to occur when the cohesion between the grains is weaker than the grains themselves, since failure in both cases occurs without deformation at the grain boundaries.

Shrinkage cavities are voids which sometimes appear in the last freezing portion of castings. Ordinary 0.35 per cent carbon steel, for example, freezes in the temperature range 2625°F to 2700°F, with a volume contraction of 3.05 per cent (A4). If the casting is not fed from some molten reservoir, a cavity will be left as solidification proceeds from the surface inwardly. An example of a shrinkage cavity is shown in Figure 3.

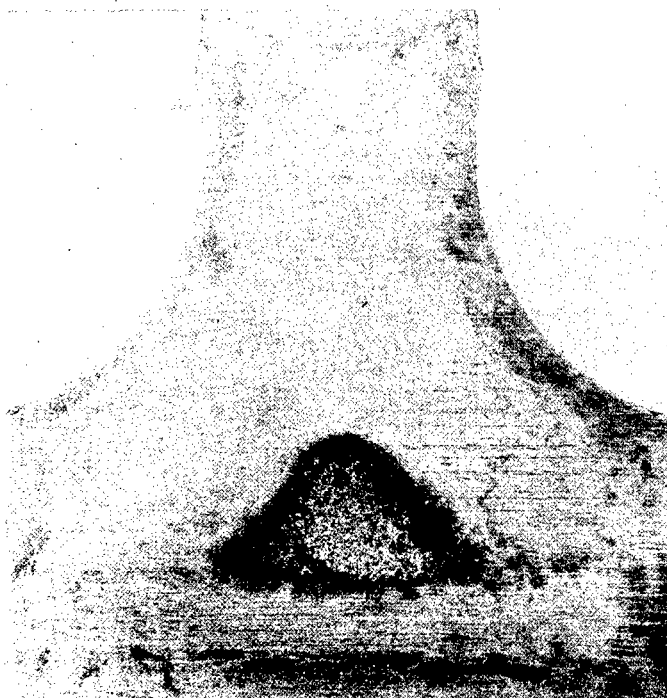
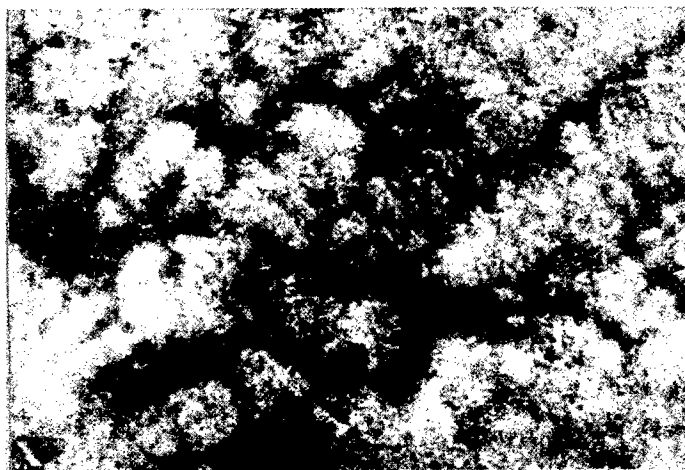


FIGURE 3. SHRINKAGE CAVITY IN A CASTING (A4)

Shrinkage cavities are commonly referred to as "pipes", such as are often observed in ingots. These are largely controlled by using hot tops<sup>(A1)</sup>.

Microscopic shrinkage cavities sometimes appear in castings at the point where dendrites, freezing from opposite directions, intersect. The contraction at the final phase of solidification produces the microscopic cavities. Portevin<sup>(A28)</sup> described them as interdendritic pipes which were formed either by the retreating flow of liquid through the entangled dendrites or by the contractions on freezing of the last portion of the liquid phase. A typical example of a microscopic shrinkage cavity is shown in Figure 4. This type of defect is commonly referred to as centerline shrinkage.



150x

FIGURE 4. MICROSCOPIC SHRINKAGE IN A STEEL CASTING<sup>(A28)</sup>

Shrinkage cavities are a result of poor design and are almost entirely free from the influence of alloying elements in the cast steel. They occur mostly in sections of large mass where hot spots are prevalent. The hot spots can be largely controlled through proper feeding of liquid metal into the points of greatest mass. The use of chills is also a very effective means of preventing the formation of hot spots. Brinson and Duma<sup>(A7)</sup> made an extensive investigation into methods of preventing centerline shrinkage cavities in castings. They found that a carefully controlled system of padding eliminated the shrinkage defects. Padding consists of tapering the walls of castings with the cross section increasing in the direction of the feed heads. There is a similarity between shrinkage cavities in castings and interdendritic shrinkage cavities in large weld-metal deposits.

## Hot Shortness

"Hot shortness" is defined as high-temperature brittleness in metals (A1). For steel, the term may be restricted to brittleness above 2000°F (sulfur) and "red shortness" used for brittleness below 2000°F (oxygen). A steel that is hot short will generally crack or break up if forged or rolled or otherwise strained in the temperature range 1800° to 2400°F. This temperature range may be wider or narrower, depending upon the composition of the steel.

Hot shortness, as distinguished from hot tearing, is an inherent weakness in some metals which makes them subject to failure when worked at high temperatures. Not all metals are hot short, whereas all metals will hot tear if strained sufficiently during solidification. For example, a pure metal, such as gold, may not be hot short but it will hot tear.

The exact mechanism of hot shortness in steel is not completely understood. Most investigators seem to agree that it is caused by certain impurities, such as compounds of sulfur and oxygen, in the grain boundary, and yet this is debatable. Although not based on actual experimental evidence, many investigators hold to the idea that hot shortness is caused by the presence of liquid eutectic films which form continuous envelopes around the metal grains at elevated temperatures. It was believed that such grain-boundary films melt at the hot-shortness temperature and thus lose strength<sup>(H211)</sup>. While that is an easy explanation, it is hardly consistent with direct observation.

A difficulty with the eutectic-film idea is that it does not clearly explain why some steels can be forged successfully when heated to temperatures above the hot-shortness range. One possible explanation for this is that the grain-boundary eutectic (FeS in this example) collected in harmless globules at higher temperatures, instead of remaining dispersed as an intergranular film<sup>(H117)</sup>. There also is the possibility that the grain-boundary materials are absorbed by diffusion or solution into the grains at the higher temperatures. Kenyon<sup>(B16)</sup> hypothesized that Armco iron loses its brittleness on heating above 1950°F by a diffusion of [FeS or FeS] eutectic throughout the entire structure at the higher temperatures. This hypothesis will be discussed later.

In partial support of this theory, Sims and Lillieqvist<sup>(B22)</sup> showed that the room-temperature ductility of cast steel was lowered by the presence of a continuous network of eutectic in the grain boundary. The eutectic appeared to be brittle and, when plastic deformation occurred, it cracked before the metal within had a chance to flow. A cast metal with high ductility contained inclusions which were globular in shape and were located mostly within the grains.

Hultgren<sup>(B12)</sup> did not find any connection between the paths of hot-shortness cracks and inclusions in steel, as determined by hot bend tests of several heats of steel which were definitely hot short. He also stated that in soft steel, hot-shortness cracks probably are always formed along the grain boundaries of ferrite below the  $A_3$  critical temperature, and of austenite above the  $A_3$  critical temperature. He further stated that, in medium-carbon steel, hot-shortness cracks sometimes follow the grain boundaries of the austenite and sometimes do not.

In contrast to the liquid-film idea, another investigator<sup>(B18)</sup> theorized that hot shortness was caused by solid grain-boundary films which were brittle and unworkable in the hot-shortness temperature range, but at higher temperatures became plastic and workable. He illustrated this by forging two halves of an ingot containing 0.07% C, 0.13% Si, 0.42% Mn, 0.042% P, 0.18% S, and 0.017%  $O_2$  at 1920°F and 2140°F, respectively. The ingot cracked badly when forged at the lower temperature, but at 2140°F it was forged successfully. Feild<sup>(B7)</sup> supported the plastic-film idea by citing the example of elongated inclusions in hot-rolled steel. He stated that the inclusions must have been plastic at rolling temperatures. He apparently reached the conclusion that inclusions and grain-boundary films were similar materials.

Hot shortness of ingot iron in the temperature range 1470°F to 1650°F was attributed to the existence of a eutectoid phase in the grain boundary<sup>(B4)</sup>. This conclusion was based on microscopic studies of specimens of ingot iron quenched from various temperatures. The occurrence and disappearance of the eutectoid phase coincided with the beginning and end of the brittle zone, as observed in practice. This investigator believed that the phase was composed of impurities such as phosphides, sulfides, iron carbide, and possibly traces of manganese sulfide and iron oxide.

Most investigators believe that sulfur is the most common cause of hot shortness in steel. The detrimental effect of sulfur is most often attributed to the formation of FeS in the grain boundaries. FeS melts at about 1850°F<sup>(H117, A1)</sup>. The microstructure of a high-sulfur (0.53 per cent) steel is shown in Figure 5. The constituent in the grain boundary was reported to be FeS<sup>(B21)</sup>. Kenyon<sup>(B16)</sup> attributed hot shortness in Armco ingot iron to a network of FeS in the grain boundaries. He based this conclusion on microscopic examination of higher sulfur samples, since the grain-boundary material could not be seen in Armco iron. He said that the FeS eutectic undoubtedly loses its strength below its melting point (about 1805°F), since the "critical working range" of Armco ingot iron is about 1650°F. It would appear that the sulfur content of Armco ingot iron (about 0.025 per cent S) is too low to produce hot shortness of the type ascribed to sulfur that comes at 2200°F. On the other hand, Armco iron is hot short at 1600°F to 1650°F at a temperature range generally ascribed to oxygen. Ihrig<sup>(B13)</sup> made a series of hot-twist tests to determine the effects of various elements on the hot workability of steel. He obtained some very interesting results which were plotted in curves showing the number of

twists required to cause failure as related to temperature. A group of these curves showing the effect of increasing sulfur content on decreasing the hot workability of steel is shown in Figure 6. The chemical compositions of the steels tested in this group are shown in Table 2.

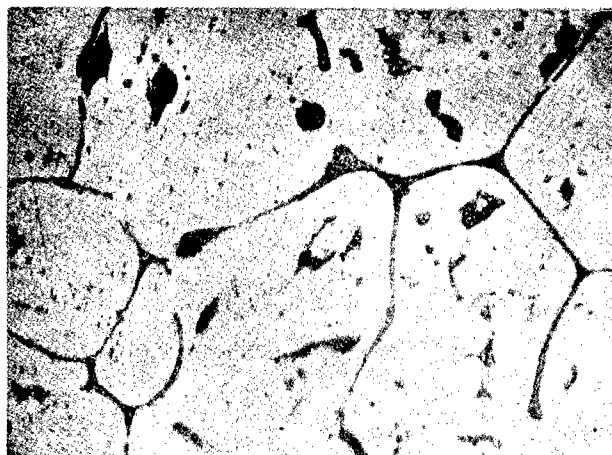


FIGURE 5. MICROSTRUCTURE OF 0.53 PER CENT SULFUR STEEL SHOWING GRAIN-BOUNDARY CONSTITUENT REPORTED TO BE IRON SULFIDE<sup>(B21)</sup>

TABLE 2. CHEMICAL COMPOSITION OF A SERIES OF INGOTS CAST FROM A SINGLE HEAT OF SAE 1020 STEEL USED IN TESTS TO DETERMINE THE EFFECT OF SULFUR ON HOT WORKABILITY<sup>(B13)</sup>

Heat No.	Chemical Composition, per cent				
	C	Mn	S	P	Si
X-518	0.22	0.45	0.021	0.014	0.08
X-519	0.22	0.48	0.052	0.014	0.08
X-520	0.23	0.48	0.070	0.014	0.08
X-521	0.23	0.46	0.048	0.014	0.08
X-522	0.24	0.46	0.070	0.014	0.08
X-523	0.24	0.46	0.116	0.014	0.08
X-524	0.24	0.44	0.130	0.014	0.08

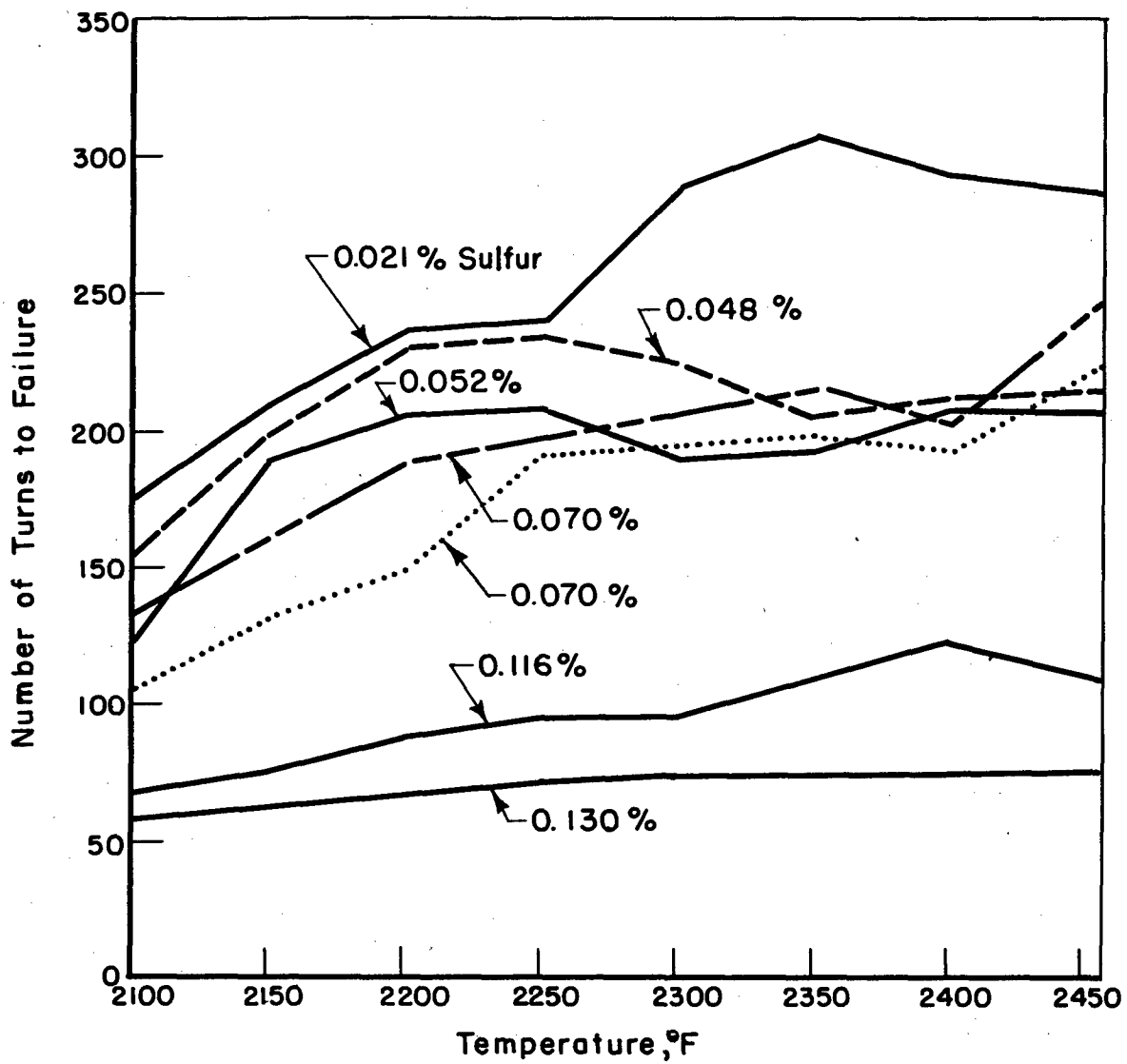


FIGURE 6. CURVES SHOWING THE EFFECT OF SULFUR CONTENT ON THE HOT WORKABILITY OF SAE 1020 STEEL AS DETERMINED BY HOT-TWIST TESTS (B13)

Sulfur in excess of 0.01 per cent in ferritic steel will cause hot shortness if manganese is not present. In most steel specifications, sulfur is usually kept below 0.05 per cent, with enough manganese added to form  $MnS$  (H15).

Oxygen is also believed to be a strong factor in causing hot shortness in metals (H104, H132, B19, B21, C36). Pure iron may become hot short if it is oxidized during melting (H132). Austin (B2) found that the hot-shortness temperature of pure iron in the region of 1620°F was lowered by adding oxygen.

Iron that has been raised to near its fusion temperature in air and slowly cooled was designated in earlier times as "burned" or overheated metal. Such iron may be hot short due to oxygen, but this oxygen is not, as is commonly believed, absorbed during heating. According to Ledebur (C36), the oxygen was previously contained within the iron itself through the medium of slag impurities or inclusions. He further states that the slag impurities are reduced by the iron to another form which causes hot shortness. Another investigator (B23) found that oxidation during heating caused hot shortness in 0.10 to 0.20 per cent carbon steel. Brittleness was found to increase with carbon content and was influenced somewhat by the Cu and S contents.

Check cracking during the tube piercing of 18 Cr - 8 Ni stainless steel billets was believed to be caused by oxygen, possibly in combination with chromium (B10). The inclusions in these steels were not excessive, but it was believed that an oxygen-bearing constituent must have formed at the piercing temperature, 2150°F to 2350°F, causing the low ductility. With ferritic stainless steels, it was believed that a direct correlation was shown between checking and oxygen content.

Hot shortness of Armco ingot iron has sometimes been blamed on oxygen because it is a relatively pure iron. However, several investigators (B14, B16) concluded that it was due to the lack of sufficient manganese to counteract the effect of sulfur.

In some instances, two samples taken from basic open-hearth charges at different stages in the course of a heat had practically identical contents of sulfur and manganese, and yet one showed marked red shortness and the other an entire absence of this condition (B19). From this, it appeared that sulfur, in itself, was not a cause of the hot shortness, while other work has shown equally that excess oxides in pure steel did not lead to hot shortness above 1650°F. It seemed to be their combined presence that was so detrimental (B19).

Niedenthal (B18) made hot impact and hot bend tests on a series of low-carbon steel melts to determine the individual and combined effects of sulfur and oxygen on hot shortness. The chemical analyses and hot-shortness temperatures are shown in Table 3. Curves showing the relative



TABLE 3. CHEMICAL COMPOSITION AND HOT-SHORTNESS TEMPERATURE OF LOW-CARBON STEEL AS DETERMINED BY HOT IMPACT TESTS(BI8)

Melt No.	Chemical Composition, per cent								Hot-Shortness Temperature, F	Remarks
	C	Si	Mn	P	S	Al	O			
2	0.16	0.15	0.41	0.01	0.15	0.04	0.19		2300	High sulfur
3	0.05	trace	0.12	0.01	0.021	0.05	0.008		None	Low sulfur, low oxygen
5	0.12	0.13	0.47	0.022	0.125	0.04	0.015		2430	High sulfur
6	0.11	0.11	0.47	0.023	0.185	0.04	0.016		2370	High sulfur
7	0.10	0.12	0.46	0.022	0.20	0.05	0.023		2370	High sulfur
11	0.11	0.13	0.46	0.021	0.26	0.04	0.014		2100	High sulfur
17	0.03	trace	0.10	0.01	0.022	(1)	0.092		1615-2100	High oxygen
21	0.03	trace	0.43	0.01	0.023	0.05	0.125		None	High oxygen

(1) Not determined.

effects of sulfur and oxygen on the hot shortness, as determined by hot impact tests, are shown in Figure 7. Note that hot shortness associated with sulfur seems to occur at a higher temperature (about 2300°F) than that associated with oxygen. Also, it was significant that the hot shortness caused by sulfur is considerably higher than the reported melting point of FeS (1850°F). This would indicate that the liquid-eutectic-film idea may not be correctly applied to the hot-shortness phenomenon. However, this conclusion was based on hot-impact-test results. Some investigators (B16) do not believe that the hot impact test is a satisfactory means of measuring hot shortness.

The combination of sulfur and oxygen seems to have an even greater effect than either of the two alone because of the formation of the eutectic compound FeO-FeS (H132, B19).

Norris (B19) cast a series of steel samples, starting after the slag was shaped up on heats of ingot iron and continuing at intervals of about an hour until the heats were tapped. The analyses of these samples and the results of hot bend tests from two of these heats are shown in Table 4. In a heat of ingot iron which is normally hot short, manganese is the last element to be eliminated. Therefore, samples taken early in the heats were not hot short. Manganese, sulfur, and oxygen appear to be the important variables in influencing hot shortness in the steel samples.

Note, in Table 4, that the start of hot shortness in Heat B was not accompanied by an appreciable change in analysis, except for a slight decrease in manganese content. The first hot-short sample in Heat A showed an increase in oxygen content and a slight decrease in manganese content. These results tend to illustrate that a very slight change in composition can make the difference between no hot shortness and hot shortness. It should be pointed out, however, that other impurities besides those analyzed for in the steel samples may have had an effect.

One investigator (H51) suggests that hot shortness, particularly in weld metal, might be caused by the formation of the compound FeO·MnO·FeS·MnS which melts at 1670°F, according to the phase diagram for this system, as shown in Figure 8.

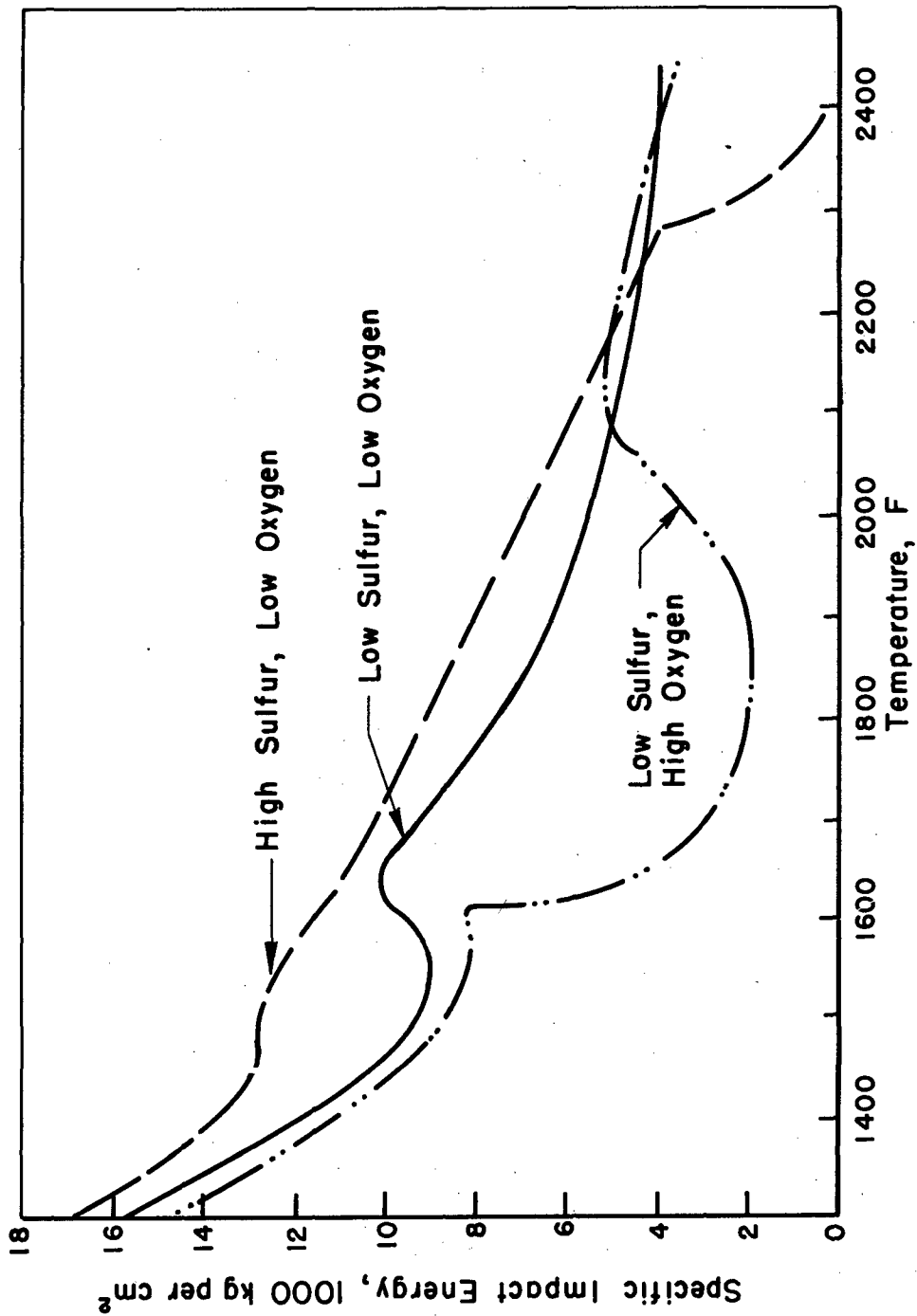


FIGURE 7. CURVES SHOWING THE EFFECTS OF SULFUR AND OXYGEN ON THE HIGH - TEMPERATURE IMPACT ENERGY OF LOW - CARBON STEEL (B19)

TABLE 4. HOT-SHORT CONDITION AND COMPOSITION OF STEEL  
 SAMPLES TAKEN FROM TWO OPEN-HEARTH HEATS<sup>(B19)</sup>

Heat	Sample	Hot Short	Chemical Composition, per cent					
			C	Mn	P	S	Si	O <sub>2</sub>
A	1	No	0.03	0.113	0.044	0.038	0.046	0.021
	2	No	0.03	0.114	0.030	0.041	0.006	0.060
	3	No	0.029	0.063	0.004	0.029	0.002	0.034
	4	Yes	0.025	0.054	0.002	0.023	0.003	0.063
	5	Yes	0.023	0.050	0.002	0.023	0.002	0.055
	6	Yes	-(1)	-	-	-	-	-
	7	Yes	-	-	-	-	-	-
B	1	No	-(1)	-	-	-	-	-
	2	No	-	-	-	-	-	-
	3	No	0.07	0.084	0.005	0.024	0.004	0.024
	4	No	0.020	0.053	0.002	0.022	0.002	0.059
	5	Yes	0.019	0.044	0.002	0.022	0.002	0.058
	6	Yes	0.017	0.027	0.002	0.022	0.002	0.028
	7	Yes	-	-	-	-	-	-
	8	Yes	-	-	-	-	-	-

(1) Not analyzed, since the borderline samples between no hot shortness and hot shortness were the only important samples.

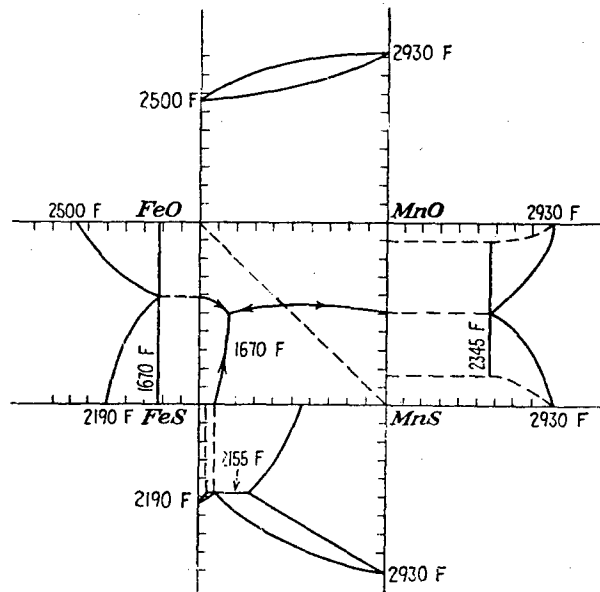
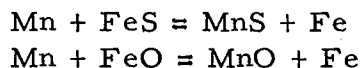


FIGURE 8. PHASE DIAGRAM MnO-MnS-FeO-FeS (A8)

Other elements in ferritic steels which have been variously related to hot shortness are: boron<sup>(H55, B24)</sup>, tin<sup>(B13)</sup>, lead<sup>(B13)</sup>, arsenic<sup>(B11)</sup>, selenium<sup>(B13)</sup>, nitrogen<sup>(B13)</sup>, copper<sup>(B10)</sup>, and columbium<sup>(H164, H51, B13)</sup>. High-temperature bend tests made by Cain<sup>(B5)</sup> did not seem to indicate that copper by itself causes hot shortness in steel. Phosphorus has long been known to have an adverse effect on the room-temperature ductility of steel, but tests by Ihrig<sup>(B13)</sup> did not indicate that it was detrimental to the hot workability of low-carbon steel.

Grain size has a secondary influence on hot shortness in steel<sup>(B19)</sup>. This is probably because the areas between the grains are wider in large-grain-size steels, which would exaggerate the effects of weak grain boundaries.

Manganese appears to be the greatest boon to metallurgists in preventing hot shortness in ferritic steels. The function of manganese is not definitely known, but it is believed to have a beneficial effect by causing the following reactions to proceed to the right<sup>(B19)</sup>:



The ratio of manganese to sulfur to prevent hot shortness was estimated to be about 3<sup>(B7, B19)</sup>. Other investigators have stated that, to be on the safe side in commercial practice, the manganese-sulfur ratio should be somewhere between four and eight<sup>(B7)</sup>.

Thermochemical evidence seems to indicate that sulfur can exist in steel as FeS and can be changed to MnS during rolling<sup>(B9)</sup>. Whiteley<sup>(B26)</sup> suggests that the amount of manganese required to convert all of the sulfur to MnS increases with the rate of cooling down to about 1650°F. This might indicate that the manganese content of weld metal should be much higher than that used in steelmaking practice because of the high cooling rates involved.

Zirconium has been used successfully as a substitute for manganese in combating hot shortness in ferritic steels<sup>(B1, B7)</sup>. Zirconium eliminates hot shortness when present in the finished steel in the proportion of 1.41 parts or more of zirconium to 1 part of sulfur; the ratio of 1.41 corresponded to the formation of the normal zirconium sulfide, ZrS<sub>2</sub><sup>(B7)</sup>. The zirconium is preferably added in the form of silicon-zirconium alloy.

The test used most often to determine the hot-shortness properties of a steel is the bend test<sup>(B12, B19, H131)</sup>. This test consists simply of heating a bar to above 2010°F and then bending it repeatedly over an anvil until cooled to about 1470°F. The hot-short fracture is sharp and well defined at right angles to the axis of the bar.

Other methods that have been used to evaluate the hot-shortness tendencies of steel are hot impact tests<sup>(B18)</sup>, upset or pancake tests<sup>(H132)</sup>, hot twist tests<sup>(B4, B6, B8, B13)</sup>, and hot tension tests<sup>(151)</sup>. One investigator<sup>(H104)</sup> used a cup test similar to that used in tests on sheet metal for evaluating hot shortness of welded joints.

Hot shortness in weld metal and its relation to weld-metal cracking is discussed at greater length later in this report under the section on metallurgical causes of weld-metal cracking.

### Overheating

Overheating is a severe grain coarsening produced by heating steel above its normal forging temperature, which varies from steel to steel. Overheating is an inherent characteristic of the steel, which can be completely independent of furnace atmosphere. The term "burnt" has been used, and still is to some extent in the steel industry, to describe an extreme condition of overheating usually held to be due to partial fusion. However, the term is misleading, since it has been shown that oxygen is not necessary to effect overheating.

Burnt steel is brittle and may fracture or fly to pieces during forging in a manner similar to a hot-short condition. The steel also has inferior properties at room temperature. It will have a low tensile strength and usually will exhibit a coarse intergranular fracture.

Mildly overheated steel can generally be restored completely by an annealing treatment (C55, C66). Heyer (C28) found that a short anneal of about  $\frac{1}{2}$  hour at 1650°F was satisfactory for low-carbon steel. However, a badly burnt steel cannot be restored by heat treatment (C52).

Stansfield (C62) was probably the first investigator to suggest that overheating was due largely to solidus melting in steel. He also stated that oxygen had an indirect effect on overheating.

Structures in steel caused by overheating can be retained for room-temperature study by quenching from the heating temperature (C60).

Howe (C30) stated that extreme overheating was a mechanical separation of the grains. He felt that the violent rifting of grains was due to the evolution of gases such as carbon monoxide, partially occluded hydrogen, and nitrogen. Stead (C64) found that in overheated steel the phosphorus had become segregated in minute globules in the grain in continuous envelopes around the grains. His tests showed that the penetration of oxygen into the steel had nothing to do with the phenomenon. Cathcart (C14) distinguished three types of overheated steel: (1) oxidized during heating; (2) oxidized during mechanical working at temperature above that of incipient fusion; and (3) heated to above the point of incipient fusion without oxidation. The first case was actual burning or combustion; the whole material would burn away if heating was continued. In the second case, the steel will crack up completely, and in the third case the steel can be restored by heat treatment.

Andrew (C7) expressed the view that overheating in mild steel occurred only when the temperature was sufficiently high to bring about a dissociation and diffusion of the carbide phase in the grain boundary. Brearley (C10) defined overheating as grain growth in steel caused by exposure to high temperatures; and "burning" as partial grain-boundary melting with or without oxidation.

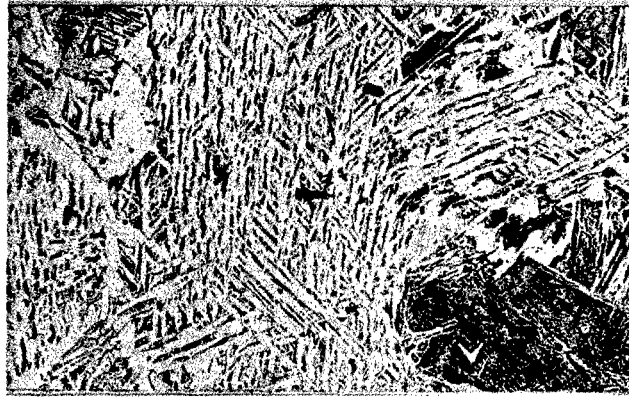
Jominy (C33) made an extended investigation into the overheating phenomenon in which he concluded that a steel was "burnt" when it contained voids around the grain boundaries, as shown in Figure 9B. He also observed that oxygen was not a necessary factor, but in most cases, burning temperatures were slightly lower in an oxidizing atmosphere. He heated steels of various compositions in reducing and oxidizing atmospheres and then forged them immediately by the upset method. The samples were then sectioned and examined under a microscope. The results of his tests are tabulated in Table 5.

TABLE 5. COMPOSITION AND BURNING TEMPERATURES OF VARIOUS  
STEELS AS DEPENDENT ON THE FURNACE ATMOSPHERE<sup>(C33)</sup>

SAE No.	Composition, per cent			Atmosphere <sup>(1)</sup>	Highest Temperature to Which Steel Was Heated Without Burning, F	Lowest Temperature to Which Steel Was Found to Burn, F
	C	Ni	Cr			
1025	0.28	0.00	0.00	Oxidizing	2545	2560
				Reducing	2625	2675
1045	0.47	0.00	0.00	Oxidizing	2475	2500
				Reducing	2575	2625
1090	0.93	0.00	0.00	Oxidizing	2400	2425
				Reducing	2400	2425
1120	1.19	0.00	0.00	Oxidizing	2320	2350
				Reducing	2320	2350
2320	0.24	3.48	0.05	Oxidizing	2570	2590
				Reducing	2600	2625
3140	0.43	1.20	0.39	Oxidizing	2500	2525
				Reducing	2570	2590
3250	0.52	2.05	0.83	Oxidizing	2500	2530
				Reducing	2525	2550

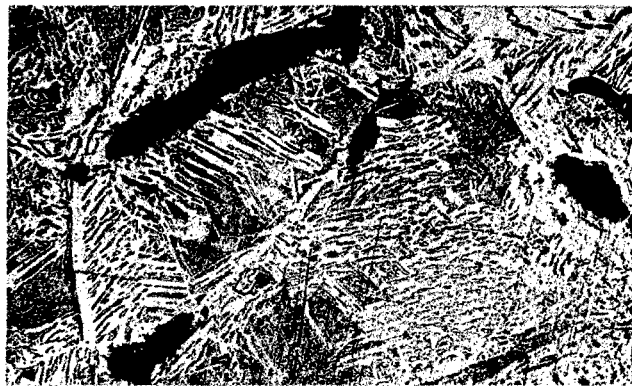
(1) In each case, the atmosphere was in a turbulent condition.





100X Nital Etch

A. Heated to 2700°F in reducing Atmosphere



100X Nital Etch

B. Heated to 2600°F in Oxidizing Atmosphere

FIGURE 9. MICROSTRUCTURES OF OVERHEATED 0.22 PER CENT CARBON STEEL SHOWING THE EFFECT OF ATMOSPHERE ON THE OCCURRENCE OF VOIDS<sup>(C33)</sup>

Herty<sup>(C27)</sup> found that the coarse grain growth in mildly overheated steel could be restored much more readily by heat treatment if the steel had been deoxidized with excess aluminum. Silicon-killed steel showed only partial recovery on renormalizing. He suggested an addition of 0.03 per cent aluminum for optimum recovery of steel from the effects of overheating. Zeigler<sup>(C77)</sup> found that 0.025 to 0.25 per cent aluminum increased the temperature at which grain coarsening or overheating in ferritic steel occurred. Colombier<sup>(C15)</sup> found that 0.10 per cent aluminum made the recovery of chromium-nickel steel from overheating more difficult.

Swinden and Sutton<sup>(C68)</sup> investigated burning of 3/32-inch-thick mild-steel (0.20 to 0.25 per cent carbon) sheet during oxyacetylene welding. The burning was developed on the underside of certain welded steel sheets when complete penetration occurred. Their results indicated that oxygen, hydrogen, and nitrogen were not associated with the burning. Residual elements in the steel appeared to increase the tendency to burning, but silicon decreased it.

Winterton<sup>(C74)</sup> investigated the effect of overheating on the transformation characteristics of a nickel-chromium-molybdenum steel (0.31% C, 0.20% Si, 0.44% Mn, 0.036% P, 0.033% S, 3.40% Ni, 1.26% Cr, 0.47% Mo). He stated that "overheating increased the inherent stability of austenite, by producing a more orderly lattice". No clear understanding of this term was offered. His results indicated that weakening in steel due to overheating did not effect a change in the dilatation characteristics of the steel. This study was made in an attempt to explain why underbead cracking in welds occurred in the overheated zone of the base metal.

Preece, et al<sup>(C46)</sup>, stated that "overheating results from diffusion processes within the austenite grain, which operate during cooling through the overheating range". They concluded that overheating was not simply the result of a mechanical envelopment of the austenite grains by the fusion of nonmetallic constituents, nor was it caused by films produced by preferential oxidation of alloying elements such as chromium.

Woolman and Kirkby<sup>(C75)</sup> concluded that faceted (intergranular) fractures of overheated steel were due to precipitation of an unknown constituent at the grain boundaries and that the constituent was already present in steels as cast. They believed that reheating to excessively high temperatures caused the critical concentration of this constituent to be exceeded.

A number of investigators<sup>(C16, C24, C34, C46, C47)</sup> have concluded that a rapid rate of cooling from the overheating temperature range caused intergranular fractures to occur during impact or tensile testing. However, the rate of cooling was critical, since very slow cooling caused intergranular fractures also<sup>(C75)</sup>. This could be an important conclusion as far as welding is concerned. Goodrich<sup>(C24)</sup> showed that slow cooling rates appreciably reduced the intergranular-fracture appearance of nickel-chromium-molybdenum-vanadium steel forgings after being heated in the temperature range 2280°F to 2550°F. He showed that a repetition of the normal oil-hardening and tempering treatments, or slow cooling from the overheating temperature, could entirely eliminate the intergranular type of fracture even in steels heated as high as 2550°F.

Thompson and Stanton<sup>(C71)</sup> made some fairly recent studies into the possible effects of oxygen on the isothermal transformation of steel. They found that the S-curve of a plain carbon (0.82% C, 0.23% Si, 0.32% Mn,

0.027% S, 0.011% P) steel was shifted to the left. They concluded that this behavior was due to the introduction of oxygen during burning.

Several investigators (H19, H92) have suggested that oxygen contamination was a cause of poor strength in pressure welds. Austin and Jeffries (H19) found that pressure welds made in hydrogen were considerably better than those made in air. The condition in as-welded pressure welds seemed to be analagous to that in wrought steel, since the fracture strength can be improved by quenching and tempering from the welding temperature, and also poor-quality welds often contain voids.

The strength and ductility of high-tensile alloy steel of low-to-medium carbon content fell off rapidly after heating in the region of 2500°F to 2550°F, and evidence of burning (solidus melting) was found after treatment above 2550°F (C49). Since the overheating temperatures of these steels were appreciably lower, it was concluded that the onset of burning was responsible for the loss of strength.

Preece and Nutting (C48) found that a saturated aqueous solution of ammonium nitrate was an ideal reagent for detecting overheating and burning in steel. The etch, used electrolytically, revealed white grain boundaries in overheated steel and black grain boundaries in "burned" steel. Hall's etch, which is composed of an alkaline solution of potassium permanganate and potassium dichromate, is very useful in detecting oxidation in steel and ferritic weld deposits.

Ko and Hanson (C34) found that a precipitation of sulfide inclusions into the grain boundary occurred in plain carbon steel when overheated. Even low-sulfur steels showed iron-rich sulfide inclusions in the grain boundary as one constituent of a eutectic. Segregation of phosphorus to the grain boundaries also occurred when iron-rich sulfide inclusions were present. Slow cooling through the overheating range reduced the number of inclusions and removed the overheating effect.

A new hypothesis has been advanced (C57) that manganese sulfide largely controls the minimum overheating temperature of steel. The temperature corresponds to some temperature at which the MnS goes into solution. According to the hypothesis, a slow rate of cooling would enable the sulfur to combine with manganese atoms in the vicinity to form minute particles of MnS, the greatest quantity being in the band around the grain boundaries. These particles effect cohesion of the grains. Extremely slow rates of cooling, however, would produce fewer nuclei and also cause coagulation of the particles. The larger inclusions have less influence on the properties of the steel. Repeated normalizing also would cause this coagulation, which explains the renewal of overheating effects by heat treatment.

## Blue Brittleness and Temper Brittleness

Blue brittleness is the reduced ductility which occurs in some steels when worked in the temperature range 300°F to 700°F<sup>(A1)</sup>. The phenomenon may be observed at the working temperature or after cooling to room temperature. Increasing the strain rate has the effect of raising the temperature range at which blue brittleness occurs. At extremely high strain rates, the blue-brittleness effect has been reported to occur as high as 1100°F<sup>(D42)</sup>. One investigator<sup>(D41)</sup> observed a brittle range for chromium-nickel-molybdenum steels which was limited to 840°F to 930°F. No brittle range was found in tests with chromium-molybdenum steels which were used as a replacement.

It has been shown that killed steels which are free from blue-brittleness characteristics are likewise free from strain-aging tendencies, thus indicating a possible common cause for these two phenomena<sup>(E10, E12, H83)</sup>. Studies have indicated that carbon and nitrogen contribute to both strain-aging embrittlement and blue brittleness<sup>(H102)</sup>.

The effect of blue brittleness was believed to be due to the precipitation of a new grain-boundary constituent. One investigator<sup>(D51)</sup> suspected the precipitation of nitrides of chromium and manganese to cause brittleness in steels annealed in the temperature range 480°F to 750°F. An addition of 0.04 per cent aluminum to low-carbon, chromium-manganese, and manganese steels greatly reduced this embrittling effect. Additions of titanium (about 5 times the carbon content) render the steel completely nonaging.

Tensile tests were made of 0.006 to 0.15 per cent C steels in the temperature range 68°F to 660°F as part of an investigation on blue brittleness<sup>(D9)</sup>. Blue brittleness was attributed to the presence of nitrogen in the steels. Nitrogen contents of only 0.0015 per cent produced a weak brittleness irregularity. The strongest effect was noted in the range of 0.002 to 0.005 per cent nitrogen. Various investigators have suggested that some types of cracking in ferritic weld metal might be attributed to blue brittleness. This is discussed in detail later in this report under the section on metallurgical factors affecting weld-metal cracking.

Temper brittleness is held to be due to a precipitation process which may be related to blue brittleness, but with the difference that straining is not an influencing factor<sup>(A1)</sup>. Temper brittleness is manifested by a decrease in impact resistance which occurs when a hardened medium-carbon structural steel is slow cooled from or held within a certain range of temperature below the transformation range. The embrittling reaction is reversible, and can be avoided by quenching from the tempering temperature.

The effect of temper brittleness is only manifested upon reheating after quenching. For this reason, it would not be expected to have an effect

on weld-metal cracking except that occurring during reheating of beads in multipass welds. However, other than lowering the notch-impact strength, temper brittleness apparently has no detrimental effect on steel. Since temper brittleness did not seem to be an important factor in weld-metal cracking, only a limited amount of literature on the subject was surveyed.

### Strain-Aging Embrittlement

All mild steels, when cold worked, show, in varying degrees, an increase in hardness (work hardening or strain hardening). This is accompanied by an increase in tensile strength, disappearance of the characteristic yield-point stretch and of the corresponding kink in the stress-strain diagram, a drop in elongation, and reduction in area and impact value. Strain-aging embrittlement is a general term applied to the continued change in physical properties when the cold-worked steel is allowed to stand at room temperature or is heated at a low temperature for a short time, a treatment sometimes known as artificial aging or "hot aging" (E1). It is particularly pronounced in its effect on notch toughness. A very good review of 40 references on strain aging in steel was prepared by Epstein (E9).

In some early investigations, it was concluded that oxygen was a major cause of strain aging (E5). However, recent work has indicated that carbon and nitrogen have the greatest effect on strain aging (E12), although oxygen may have a secondary influence (E10). High nitrogen and phosphorus contents were believed to be a contributing cause of strain aging in Bessemer steels (E1).

A fully killed steel is least susceptible to strain aging, which again seems to indicate that oxygen, or some compound such as FeO, may cause strain aging. However, strong deoxidizing agents are also strong nitride formers. For example, it was found that aluminum, titanium, and vanadium additions to steels will make them "nonaging". These elements are believed to counteract the effects of nitrogen. Steel treated with large amounts of titanium (about 5 times the carbon content) exhibits neither strain aging nor quench aging (A1).

Low ductility of weld metal has been attributed to strain aging. This will be discussed in detail later in this report.

## Quench Cracking

Quench cracking arises from conditions set up by temperature gradients in steel produced by rapid cooling. Because of these temperature gradients, the contraction arising from the decreasing temperature and the expansion arising from the austenite-martensite transformation do not occur uniformly over the parts, thus causing stresses of sufficient magnitude to lead to cracking.

Several investigators (F7, F11, F16, F17) have described the occurrence of microscopic quench cracks in martensite which caused brittleness in fully hardened and tempered steel. A photomicrograph of typical microcracks in martensite is shown in Figure 10.



2200X

FIGURE 10. MICROSCOPIC QUENCH CRACKS IN THE MARTENSITIC STRUCTURE OF A 1.13 PER CENT CARBON STEEL (F7)

These microcracks were generally developed by etching and, therefore, they might have been designated as "etching cracks". However, a direct correlation was found between the occurrence of these cracks and the ductility and impact strength of the steel (F7). They were thought to be true

cracks and not merely lines of strain opened up by etching. The microcracks were believed to result from the large-dimension changes, within an individual austenite grain, accompanied by a stepwise transformation to martensite. In some instances, the cracks were still present after heating martensitic specimens as high as 1800°F. Heating to 2000°F seemed to cause the cracks to weld together, thus eliminating them. The authors (F7) recommended interrupted quenching as a means of preventing the formation of the microcracks. A number of investigators (F6, F13) have discussed the advantages of austempering and martempering in preventing quench cracking. One investigator (F15) suggested that quenching cracks were due to shrinkage of a thin layer of martensite on the surface of steel. Another investigator (F23) stated that quench cracking was a result of tensile stresses, since failure by shear was impossible. Shear failures would require stresses much greater than that caused by quenching.

Jaffe and Hollomon (F10) stated that quench cracking could be prevented by increasing the hardenability of the steel without appreciably lowering the martensite temperature range. Carbon, for example, affects the susceptibility to cracking, because it lowers the Ms temperature and also decreases the ductility of martensite so that thermal stresses cannot be as readily relieved by plastic flow. The following recommendations were made for producing steels less susceptible to quench cracking:

1. Keep the carbon content as low as possible.
2. Silicon should be as high as is consistent with good steel quality and toughness.
3. Manganese, chromium, and nickel concentrations should bear a definite relation to each other and should be as high as is necessary to give the required bainitic hardenability.
4. Molybdenum should be added in the quantity necessary to give the required pearlitic hardenability, once the required bainitic hardenability has been obtained.

Quench cracking in alloy steel castings was found to be caused by alloying elements which lower the martensite transformation temperature of steel (F21). Carbon was the most powerful element in this respect. Manganese, chromium, and phosphorus were detrimental, but to a lesser degree than carbon. Sulfur, silicon, nickel, molybdenum, boron, and aluminum appeared to have little effect on quench cracking in cast steel.

A statistical investigation was made into the causes of quench cracking in low-alloy-steel gun tubes (F12). A definite correlation was found between the frequency of quench cracking and various factors such as carbon content, pouring temperature, ingot size and forging reduction, method of quenching, and finishing temperatures. Cracking increased with the carbon content

and ingot size, but decreased with increased pouring temperature, finishing temperature, and forging reduction.

Spretnak and Wells<sup>(F20)</sup> also made a statistical analysis of quench cracking in hollow cylinders of medium-alloy steel (0.25 to 0.50 per cent carbon). They found that cracking susceptibility was a heat characteristic, in that losses varied from 0 to 100 per cent, with only minor variations in chemistry from heat to heat. Carbon had the strongest over-all effect on the susceptibility to quench cracking. Cracking decreased with increased pouring temperature of the steel. They also found that cracking increased with ingot size, but decreased with forging reduction. Cracking was most prevalent in cylinders produced from bottom thirds of ingots. The cracking in this instance was believed to be related to the cone of solidification in the ingot.

Liedholm<sup>(G46)</sup> found that quench cracking in SAE 4330 steel was increased when using hydrogen atmospheres in heat treatment. This seems to suggest that hydrogen may play more than a small part in quench cracking. Zapffe<sup>(G70)</sup> also suggested that hydrogen may be a very important factor in quench cracking.

### Hydrogen Embrittlement

Hydrogen embrittlement has long been a problem with many users of steel, and, as such, has received wide coverage in the literature. As part of a fundamental research program at Battelle, Zapffe and Sims<sup>(G80, G81)</sup> made an exhaustive survey of literature on the problem prior to 1940. These same authors prepared a very good introduction to hydrogen embrittlement which was used as the basis for discussion in many papers written on the subject up to 1940<sup>(H297)</sup>. Much of the following information was obtained from this introduction.

Hydrogen has an embrittling effect on steel because of its ability to collect under tremendous pressure within discontinuities in the steel. The discontinuities may be blowholes, inclusions, macroscopic cracks, and grain boundaries, or submicroscopic rifts in the fine structure. The build-up of pressure in these discontinuities is due to the fact that hydrogen in the atomic form is highly soluble in steel at high temperatures, but only slightly soluble at room temperature. The solubility of hydrogen in pure iron is shown in Figure 11.



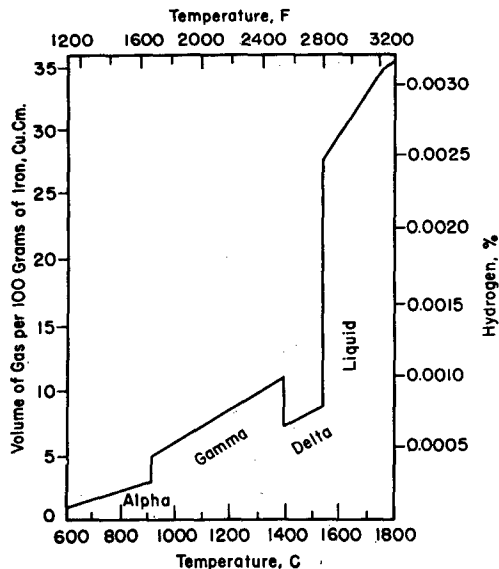
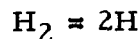


FIGURE 11. SOLUBILITY OF HYDROGEN GAS IN IRON AT 1 ATMOSPHERE PRESSURE (J47)

The solubility varies with the square root of the pressure at constant temperature following the mass-action law:



$$K_H = \frac{(PH)^2}{(PH_2)}$$

$$P_H = K_H \cdot (PH_2)^{1/2}$$

On cooling, the steel may become supersaturated with hydrogen to such a degree that the excess hydrogen starts to diffuse out of the metallic lattice. Upon reaching the internal surface of a discontinuity, the atomic hydrogen recombines to form molecular hydrogen which is extremely insoluble in steel. This reaction continues until the pressure in the discontinuity reaches equilibrium or the metal ruptures and allows the molecular hydrogen to escape.

Hydrogen apparently does not form any intermetallic compounds in steel, such as hydrides; therefore, it always behaves as a gas<sup>(G61)</sup>. It also appears to be the only substance which always behaves as a gas in steel.

Cold working appears to have a marked effect on the solubility of hydrogen in steel. The amount of hydrogen which could be introduced into Armco iron specimens under experimental conditions was only about 2 cc/100 g<sup>(G4)</sup>. Cold working the iron to 70 per cent reduction increased this saturation value to 22 cc/100 g. Cold drawing of Type 440 C steel also caused a marked increase in hydrogen absorption during pickling. The effect of cold working was believed to be due to large grains being broken

down into smaller grains (G4). This idea was based on the assumption that hydrogen occlusion occurs at the grain boundaries. Another postulation was that hydrogen operates along planes of looseness within the grain boundary (G77). It was assumed that both cold work and hydrogen act to open and develop the latent looseness structure of annealed metal, and thus their action was cooperative.

Alloy additions may increase the solid solubility of hydrogen in steel (G11) by their effect on the amount of retained austenite in the steel. This is an important consideration in underbead cracking encountered in welding hardenable steel. Underbead cracking in welds will be discussed later on in the report.

Zapffe and Sims (G82) proposed that hydrogen embrittlement was due to the occlusion of hydrogen under high pressure in "interblock disjunctions". They stated that hydrogen embrittlement was well explained by the "mosaic" block concept, for the pressure of the occluded hydrogen was aerostatic and, therefore, was exerted triaxially. Metal stressed triaxially cannot flow, and only rupture can relieve a superimposed stress.

Zapffe and Moore (G79) studied rift openings in the structure of iron containing hydrogen, and concluded that failure of hydrogen-embrittled iron occurred by direct cleavage through these gas-filled rifts, which have no ductility. They presented further evidence for the "mosaic" structure concept.

Specimens of ingot iron were charged with hydrogen and subsequently deformed under oil (G73, G77). Hydrogen was seen to evolve from deformation and slip markings only. According to the investigators, this further supported the planar-pressure theory for hydrogen embrittlement and the presence of a mosaic structure in metal. Zapffe (G74) suggested that hydrogen embrittlement could be of two types: (1) intracrystalline planar-pressure embrittlement caused by precipitation of hydrogen along crystallographic planes; and (2) intercrystalline embrittlement caused by accumulation of gaseous hydrogen reaction products in grain boundaries.

Iron and steel specimens charged with hydrogen showed a considerable decrease in brittleness when tested below  $-166^{\circ}\text{F}$  (G19). The brittleness reappeared at temperatures above this point. The explanation was based on a reduction of the internal pressure of hydrogen at the lower temperatures, resulting in decrease of triaxial stresses.

"Snowflakes", "fisheyes", and "silver streaks" in steel were believed to be caused by hydrogen embrittlement that localized around some interstice, inclusion, or blowhole (G81). The inclusion or void resembles the pupil of an eye. Fisheyes show up as small bright areas in the otherwise dull background of a freshly fractured surface (G34). Fisheyes do not exist as such in unfractured metal. Artificial aging at  $400^{\circ}\text{F}$ , or natural aging

at room temperature, prevented fisheye formation in otherwise susceptible high-tensile steel<sup>(G34)</sup>.

"Flakes" and "shatter cracks" in steel were explained as internal hydrogen-embrittled zones that cracked from stresses produced during cooling<sup>(G81, G70)</sup>. One investigator<sup>(G34)</sup> described flakes as minute, spontaneous, internal ruptures. It was believed that flakes and fisheyes were formed by the same mechanism, and that the only difference was the amount of hydrogen involved<sup>(G1)</sup>. The amount of hydrogen required for flake or shatter-crack formation appeared to be much greater than that required for mere embrittlement alone. An example of shatter cracks in rail steel that had been blown with hydrogen while molten is shown in Figure 12.



FIGURE 12. SHATTER CRACKS IN RAIL HEAD  
ATTRIBUTED TO HYDROGEN<sup>(G69)</sup>

Tears or fissures which were sometimes observed on tensile specimens after testing were also believed to be due to embrittled zones which could not accommodate plastic flow<sup>(G70)</sup>. Their formation was analogous to the formation of flakes and shatter cracks.

Flakes in steel have been traced back to melting practice where moisture in the furnace gases was believed to supply the hydrogen because of the reaction<sup>(G29, G58)</sup>:



Basic open-hearth steel was generally more crack susceptible than basic electric steel, and both types were more prone to crack because of

hydrogen embrittlement than acid open-hearth steel<sup>(G26)</sup>. The acid electric furnace appeared to be most amenable to the control of hydrogen elimination<sup>(G63)</sup>. Drying or preheating of furnace additions was suggested for preventing hydrogen absorption in the electric-arc furnace<sup>(G63)</sup>. One investigator<sup>(G53)</sup> even traced hydrogen pickup to an addition of damp coke.

Measurements of the hydrogen content of liquid steel indicated that, even when steel was melted under carefully controlled conditions, hydrogen contents ranging from 4 to 6 cc/100 g were obtained<sup>(G64, G65)</sup>. This was significant when considering that hydrogen in excess of 2 cc/100 g caused embrittlement in artificially impregnated steel. Other investigators<sup>(G63)</sup> report hydrogen in amounts of 0.20 to 0.30 RV (relative volumes) in steels of casting grades at the ladle.

The hydrogen content of liquid steel showed a rather close correlation with humidity<sup>(G53, G62)</sup>. Boiling appeared to favor the elimination of hydrogen from liquid steel in electric-arc melting<sup>(G25)</sup>. The explanation for this was that gas bubbles (CO gas) tend to sweep out the hydrogen. It was for this reason that killed steel picked up higher percentages of hydrogen than unkilld steel.

Hydrogen in steel is derived from several sources. One source is pickling acids which supply nascent hydrogen at room temperature. Noble<sup>(G51)</sup> described difficulties with hydrogen embrittlement in SAE 4640 (0.40% C, 0.70% Mn, 0.30% Si, 1.85% Ni, 0.25% Mo, 0.04% S and P maximum) steel after being pickled with 10 per cent sulfuric acid solution. He concluded that susceptibility to hydrogen in general increased with the hardness of the steel. Steels heat treated to a hardness of not over 40 Rockwell C did not exhibit hydrogen embrittlement.

Low-carbon manganese-iron alloys, containing 9 to 22 per cent manganese, were embrittled by heating in a hydrogen atmosphere at 1830°F, and then following with a water quench<sup>(G66)</sup>. Alloys containing 10 to 14 per cent manganese were embrittled by pickling in sulfuric acid. Manganese appeared to favor absorption of cathodic hydrogen during pickling.

Hydrogen is a common cause of embrittlement of wire encountered in pickling<sup>(G59, G72, G76, G84)</sup>. A brittle circumferential ring was found to progress inward in carbon-steel wire, as pickling proceeded, until the entire cross section was embrittled<sup>(G72)</sup>. Soaking in hydrogen at 2000°F produced similar embrittlement. Long-time aging (114 days) at room temperature or short-time aging (23 hours) at 225°F removed the embrittling effect.

Bend tests were used to study hydrogen embrittlement of stainless steel wire during pickling<sup>(G84)</sup>. Embrittlement occurred only with the hardenable grades, and the carbon content, and possibly small amounts of nickel, was the major contributing factor. High-carbon stainless (Type

440 C) steel wire was found to be most susceptible to hydrogen embrittlement during pickling<sup>(G76)</sup>. SAE 1020 wire was next in susceptibility, with stainless Type 431 showing only slight susceptibility.

Another source of hydrogen in steel is moisture in the atmosphere of heat-treating furnaces which might be reduced at the steel surfaces during heat treatment. Derge<sup>(G29)</sup> suggested that hydrogen pickup during the treatment of steel in soaking and heat-treating furnaces might be more important in causing hairline cracks than hydrogen pickup during electric-arc furnace melting.

Soaking of alloy steels (0.28% C, 0.26% Mn, 3.3% Ni, 1.35% Cr, 0.31% Mo) in hydrogen at 2190°F followed by a water quench resulted in the formation of hairline cracks. An example of one of these cracks is shown in Figure 13.



A. 1300X Nital Etch



B. 75X Villela's Reagent

FIGURE 13. HAIRLINE CRACKS IN QUENCHED STEEL ATTRIBUTED TO HYDROGEN<sup>(G3)</sup>

The cracks were not found in specimens examined immediately after quenching, but only after an aging period<sup>(G2)</sup>. The defects were not observed if the steel was soaked at 2190°F in a vacuum, or in nitrogen, followed by a water quench which definitely marked hydrogen as a cause. An important observation made by these investigators<sup>(G2)</sup> was that only forged steels were subject to cracking. As-cast specimens of the same steel did not crack when given the same heat treatment.

It was suggested that hydrogen might be a very important factor in quench cracking<sup>(G71)</sup>. In an investigation on quench cracking in propeller steels<sup>(G46)</sup> (similar to SAE 4330 and 4320 steels), it was found to be

definitely related to hydrogen in the furnace atmosphere. Steels heated in hydrogen cracked during quenching, whereas steels heated in a "Monogas" type atmosphere containing approximately 8.5 per cent hydrogen and a dew point of 14°F were crack free. Cracking was eliminated even when using a hydrogen atmosphere by using a quench-delay — periods up to 3 minutes — air cooling before quench.

Zapffe (G74) stated that the hydrogen potential of steam in contact with steel at elevated temperatures exceeded the potential of pure hydrogen itself. Thus, steel could dissolve hydrogen at a greater rate in a furnace atmosphere containing moisture than in one containing pure dry hydrogen.

Another investigator (G24), however, found that hydrogen in reheating furnace atmosphere had no apparent effect on the hydrogen content of steel ingots. He concluded that local concentrations of hydrogen in the atmosphere caused by the action of water vapor on iron would not likely exceed 2 per cent.

Andrew, et al (G5, G6), found that hairline cracks attributed to hydrogen in steel occurred even in the absence of transformation stresses. They stated that these stresses are only important in that they determine the particular crack distribution which results. They did not observe any cases of internal crack formation in steel in the absence of hydrogen. However, certain specimens showed no cracks, in spite of a high hydrogen content. They did not know the exact nature of the hydrogen-rich constituent in steel which caused embrittlement, but stated that it was not necessarily austenitic.

In an investigation of 22 different steels, a marked relation between hydrogen evolution and the transformation characteristics of steel was observed (G7). An increase in the rate of hydrogen evolution invariably occurred at the austenite-ferrite-transformation temperature range. Below this range, the rate of hydrogen evolution decreased with temperature.

"Pinholes" and gassiness in castings were believed to be caused by hydrogen originating from moisture in sand molds (G70, G83). The effect was felt to be caused by hydrogen in the atomic form, entering the surface of the casting and then reducing oxides in the steel. Insoluble water vapor formed by this reaction caused the porosity. An example of porosity in a steel casting caused by hydrogen is shown in Figure 14.

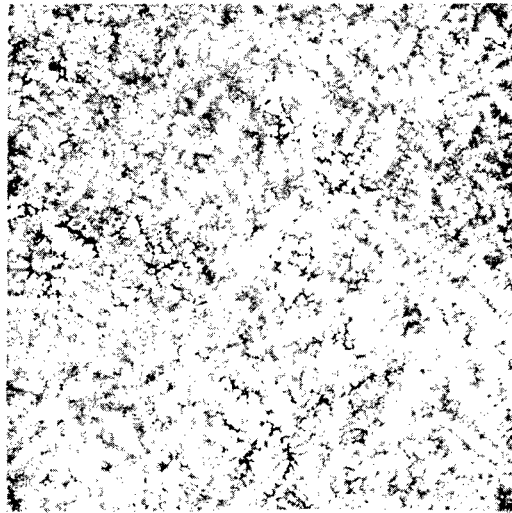


FIGURE 14. INTERDENDRITIC POROSITY IN A NICKEL-STEEL CASTING ATTRIBUTED TO HYDROGEN<sup>(G69)</sup>

This porosity resembles the interdendritic shrinkage cavities described in the section titled "Hot Tears and Shrinkage Cavities". Austenitic steels, such as high-nickel steels, were less susceptible to gassiness because the solid metal was a good solvent for hydrogen<sup>(G84)</sup>.

The ability of hydrogen to reduce ductility was demonstrated in tensile tests of cast steels containing varying amounts of hydrogen<sup>(G62)</sup>. A schematic diagram showing the trend lines for elongation values for tensile specimens from 14 heats of Grade B cast steel, as affected by hydrogen content, is shown in Figure 15.

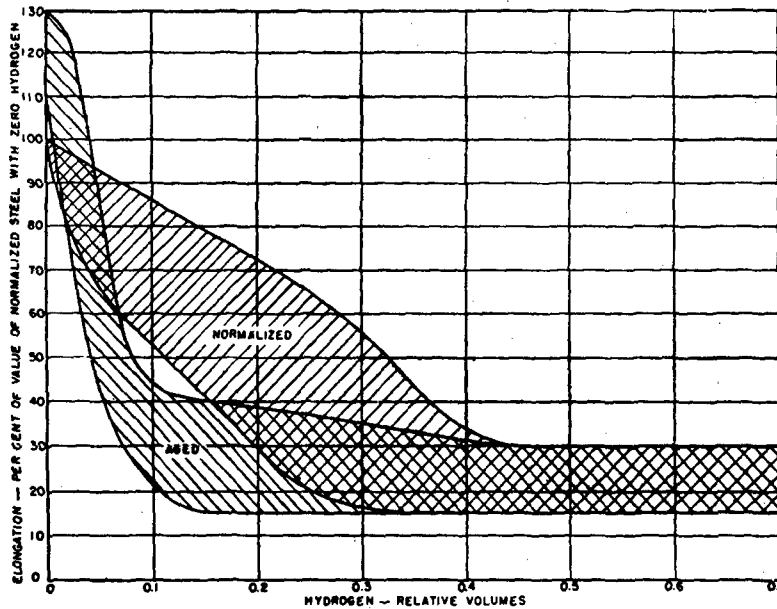


FIGURE 15. INFLUENCE OF HYDROGEN ON DUCTILITY OF CAST STEEL (G63)

The normalized specimens had been held for two hours at 1650°F and then air cooled. The aged specimens were held for 25 hours at 700°F. These curves showed that the effect of hydrogen on ductility varies with the heat treatment. Aged specimens did not show any improvement in ductility until hydrogen was less than 0.10 relative volumes. The proportional effect of hydrogen was about four times as great as in normalized specimens. The authors (G63) reasoned that, in aging, the first portion of hydrogen to diffuse out was so located as to have little or no effect on ductility.



## WELD-METAL CRACKS AND RELATED DEFECTS

In the technical literature, the most frequently mentioned weld-metal cracking is the so-called "hot" cracks. Hot cracking generally refers to failures which occur in weld metal as it freezes, or while it is still at relatively high temperatures. Many investigators believe that hot cracks are related to hot tearing in castings, although no definite proof of this has been found in the literature. The crack surfaces are usually blued by oxidation. Originally, hot cracking was meant to describe cracks which were visible on removal of the slag from the weld metal immediately after deposition. However, by general usage, the term has often been used to describe any longitudinal crack in weld metal.

Cracks which occur in weld metal at relatively low temperatures, probably below 400°F, are generally called cold cracks. Sometimes these cracks do not develop until several days after welding. Until recent years, cold cracking in the weld metal has not received much attention in the literature\*. This may have been either because cold cracking was not detected or because the more susceptible high-strength welding electrodes were not in common use. Work by several investigators (H37, H93, H214) has indicated that even low-carbon steel weld metal, such as that deposited with E6010 and E6011 electrodes, may be susceptible to cold cracking. Microfissures were observed in such weld-metal deposits if they were rapidly cooled by quenching. The microfissuring did not occur if a 300 F preheat temperature was used, or if the welds were made with low-hydrogen welding electrodes. The suggestion is that hydrogen may have caused the microfissuring, but this has not yet been definitely established. Other investigators (H264) have produced cold cracks in high-strength steel weld-metal deposits by quenching after welding. These were also believed to be caused by hydrogen.

The literature cites several instances of cracks in weld metal which do not seem to come under the classification of either hot or cold cracks. For example, some investigators (H83, H102, H271) have suggested that some types of weld-metal cracks might occur at intermediate temperatures, as a result of phase transformations, precipitation phenomena, and blue brittleness.

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\* Most references to "cold cracking" or "underbead cracking" deal with cracks in the heat-affected parent metal adjacent to welds made with high-hydrogen electrodes. These references are not of primary importance to this review, except as they describe the relation between hydrogen in metal, cooling rate after welding, steel composition and structure, and stress. For the reader's convenience, the bibliography on parent metal cold cracking is given in Section J of the bibliography.

## Classification by Microscopic Appearance

When a crack in metal occurs, its cohesive strength is less than its flow strength, or the stress pattern is such that it produces multidirectional stress. Cohesive failure may occur either through the grains or along the grain boundaries, depending upon which path is weaker. Just below the melting point of a metal, the cohesive strength of the grain boundary is substantially zero, though the cohesion within the grains could be considerable<sup>(K5)</sup>. In hot tears in steel castings, the separation follows the austenite grain boundaries for this reason. With decreasing temperature, however, and depending on the strain rate, the strength of the grain boundary increases rapidly until it equals that of the grain. This temperature is usually referred to as the equicohesive temperature. Below this temperature, the grain-boundary phase becomes stronger than the grain.

It follows that weld-metal cracking which occurs below the equicohesive temperature of steel would be expected to be transgranular rather than intergranular. However, the presence of oxides, sulfides, or other impurities in the grain boundaries changes the conditions. The cohesion is no longer between two metal grains but between a metal and a nonmetal. In such cases, intergranular cracking could occur at low temperatures.

### Intergranular

In the literature and in printed discussions of weld-metal cracking, the terms intercrystalline, intergranular, and interdendritic have been used more or less interchangeably to describe the course of a crack. There has been some confusion as to the meaning of these terms. The following definitions of a crystal, a grain, and a dendrite are set down according to the American Society for Metals Handbook:

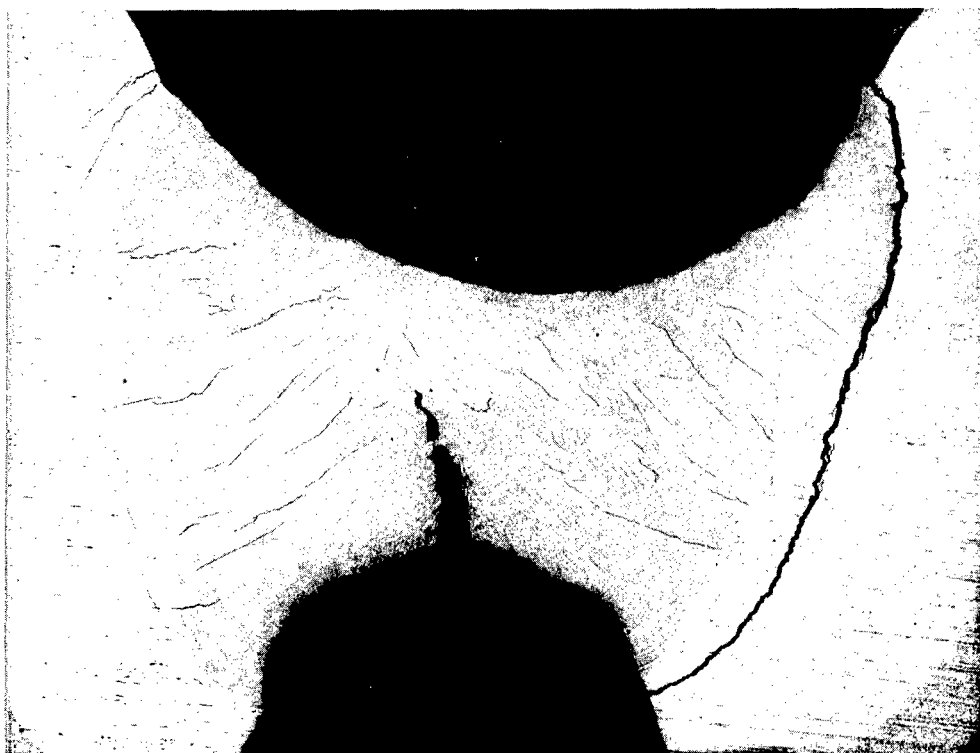
A crystal is a physically homogeneous solid in which the atoms, ions, or molecules are arranged in a three-dimensional repetitive pattern.

A grain is an individual crystal in metal.

A dendrite is a crystal formed usually by solidification and characterized by a treelike pattern composed of many branches; also termed "pine tree" and "fir tree" crystal.

In light of these definitions, a dendrite is a special type of primary crystal or grain "crystallite" which forms during the solidification of metal. In weld metal, for example, freezing of a dendrite starts at the interface between the weld metal and the base metal and proceeds toward the center of the weld. Because of the temperature gradient, molten metal precedes the

growing dendrites, and they tend to grow into the melt in a columnar fashion until they meet columnar dendrites growing from the opposite side of the weld. In Europe, these dendrites are often referred to as columnar crystals, and, in the United States, as columnar grains. For purposes of clarity in this survey, however, the term "dendrite" will be used wherever applicable. Cracking which occurs at the boundaries of these columnar dendrites may be referred to specifically as interdendritic cracking. A very good example of this type of cracking is shown in Figure 16.



8X

Aqua Regia Etch

FIGURE 16. CRACKS IN THE COLUMNAR DENDRITE BOUNDARIES OF WELD METAL DEPOSITED WITH E10020 ELECTRODES ON SAE 4340 BASE METAL

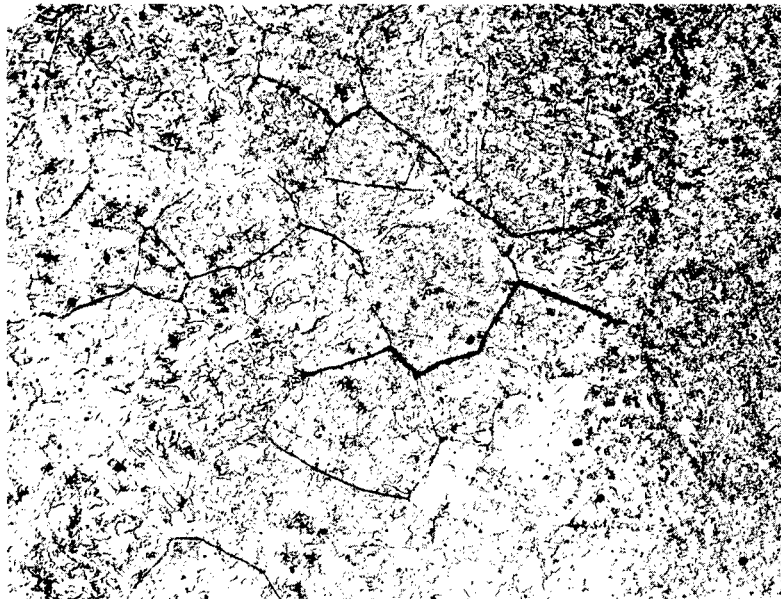
(Cracks in the base metal are underbead cracks)

When cast steel cools through the austenite temperature range, dendrites undergo a process of granulation<sup>(A4, B21)</sup>. The dendrites break up into austenite grains, each one with its own orientation, and therefore of independent crystalline existence. The dendrite disappears as a crystalline unit and the austenitic grain boundaries do not seem to bear any relation to the original dendritic grain boundaries<sup>(A4)</sup>. In an as-cast structure, the dendritic pattern, as shown in Figure 16, can usually be brought out by

etching, because of segregation which is not affected by granulation. In this survey, cracking which follows the austenitic grain boundaries (when ever established) will be described as intergranular.

Cast steels sometimes exhibit intergranular fractures when broken at room temperature. These fractures follow along what was once the primary austenite grain boundaries<sup>(A22)</sup>. The intergranular fractures in cast steels were believed to be primarily caused by aluminum nitride precipitation, ferrite precipitation, or carbide precipitation in the grain boundaries<sup>(A23)</sup>.

An interesting type of weld-metal cracking which was referred to as being intergranular is shown in Figure 17. These cracks were found in weld metal deposited by the atomic-hydrogen welding process, using SAE 6130 steel filler wire and SAE 4320 steel base metal. The welds were made under very high restraint, using 600° to 700°F preheat. The atomic-hydrogen weld was covered with a second layer of weld metal deposited with the submerged-arc welding process. After welding, the entire welded assembly was heated for 1 hour at 600°F and then stress relieved for 2 hours at 1200°F. The cracks were confined to the atomic-hydrogen weld layer and were detected after machining away the submerged-arc weld layer. The cracks followed the boundaries of large polyhedral grains which were believed to be prior austenitic grains. However, since Figure 17 shows a section parallel with the top surface of the weld, the grains could actually be the ends of dendrites.



100X

FIGURE 17. INTERGRANULAR CRACKS IN WELD METAL DEPOSITED WITH SAE 6130 FILLER ROD USING THE ATOMIC-HYDROGEN WELDING PROCESS<sup>(H225)</sup>

## Transgranular Cracks

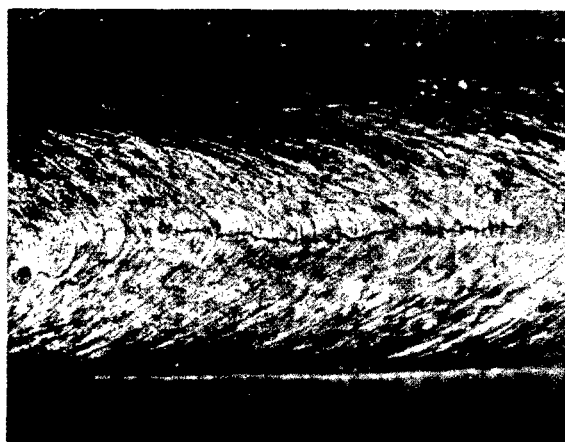
Relatively few examples of transgranular cracking in weld metal were found in the literature. One specific type of transgranular cracking, previously described, was that encountered in the reinforcing beads of weld metal deposited with stainless Type 310 (25 Cr - 20 Ni) electrodes on armor<sup>(H99)</sup>. The authors reported that these cracks were purely trans-crystalline, and that they were cold cracks occurring at a relatively low temperature. They believed that the cracking was directly caused by base-metal dilution which caused the formation of an air-hardening alloy whose suppressed transformation temperature developed sufficiently high internal stresses to produce cold cracks.

### Classification by Visual Appearance

Since the terms "hot cracks" and "cold cracks" appeared to be misleading, a classification by visual appearance has been selected for this report. It was felt that this method of classification should cover all of the known types of cracking observed in weld metal.

### Longitudinal Cracks

Various cracks commonly referred to as centerline-bead cracks, welded seam cracks, root cracks, crater cracks, and fillet cracks can be classified under the one heading "longitudinal cracks" since they all occur longitudinally to the direction of welding. They are often seen running along the top surface of a weld-bead deposit, as shown in Figure 18.



2X

FIGURE 18. LONGITUDINAL CRACK IN WELD METAL<sup>(H215)</sup>

Longitudinal cracks result when the stress transverse to the direction of welding exceeds the fracture strength of the weld metal at the temperature of cracking. Most longitudinal cracks seem to form while the weld metal is still very hot. Rollason<sup>(H210)</sup> stated, "High-tensile alloyed weld metal is subject to longitudinal cracking at the throat of the weld where columnar crystals meet to form a weak plane." A very good illustration of a centerline-bead crack in a submerged-arc weld deposit is shown in Figure 19. This crack did not open to the surface and therefore was probably similar to an internal hot tear or shrinkage cavity in a casting.

Longitudinal cracks are most common in the first pass of a multipass weld. It is generally held that, if the first pass can be deposited without cracking, the remaining passes will be crack-free. Root beads and fillet beads with a concave contour are most susceptible because of their thinner and weaker cross sections. The crater is generally the thinnest section of a weld and will often contain cracks when the remainder of the weld is crack-free. Austenitic weld metal is generally more susceptible to longitudinal cracking than is ferritic weld metal<sup>(I11)</sup>.

The so-called "interface cracks" or dilution-zone cracks occur longitudinally near the fusion line of dissimilar-weld deposits, i. e., welds deposited with austenitic electrodes on ferritic base metal. These cracks are generally limited to a narrow zone of alloy with poor properties produced by a combination of the base metal with the high-alloy weld metal. There seems to be some disagreement as to whether this zone is in the unfused base metal or in the weld metal. Linnert<sup>(I23)</sup> described these cracks as being in an unfused zone, just outside the weld zone, which had a composition different than the base metal. He attributed the alloy pickup to solid diffusion from the weld metal during welding. However, it is very unlikely that solid diffusion could occur rapidly enough in the short time at welding temperature. Other investigators<sup>(H51, H278)</sup> believe that the zone was in the fused weld metal rather than in the base metal.

Another type of "interface" cracking was encountered in welding the heat-resisting S-590 alloy with S-590 filler wire<sup>(H123, H164)</sup>. The welds were deposited in a deep V-groove joint using a ring backup, in the fabrication of gas-turbine rotor discs. The cracks followed closely along the fusion line, partly in the weld metal and partly in the base metal, as shown in Figure 20. The cracking appeared to be related to hot shortness of the S-590 alloy<sup>(H123, H164)</sup>. The high strength of the alloy at elevated temperatures was also felt to be a contributing factor because of the higher stresses which could develop during cooling. It might have been that the alloy would not yield to relieve the stress.



FIGURE 19. INTERNAL CENTERLINE CRACK IN A  
SUBMERGED-ARC WELD-METAL DEPOSIT



2X

Aqua Regia Etch

FIGURE 20. CRACK ADJACENT TO FUSION LINE IN  
S-590 ALLOY WELD METAL DEPOSITED  
ON S-590 ALLOY BASE METAL (H164)

### Transverse Cracks

Transverse cracks occur in weld metal as a result of excessive longitudinal stress. An example of a transverse crack is shown in Figure 21.





FIGURE 21. TRANSVERSE CRACKS IN WELD METAL<sup>(H216)</sup>

In some instances, transverse cracking has been related to differences in thermal expansion between the weld deposit and the base metal which developed stress during cooling after welding<sup>(H65)</sup>. This may be the reason why hard-facing deposits are prone to crack transverse to the direction of welding. For example, cracks in Stellite (50% Co, 27% Cr, 14% W, 2.6% C) hard-facing deposits on valves containing 0.4% C, 13% Cr, 13% Ni, 2.5% W were attributed to differences in thermal expansion which caused stresses to develop during cooling after welding<sup>(H65)</sup>. The Stellite had a lower coefficient of expansion ( $12 \times 10^{-6}$  cm per degree C) than the valve ( $17.5 \times 10^{-6}$  cm per degree C). Avery<sup>(H22)</sup> believed that transverse cracks in some hard-facing deposits were caused by the difference in high-temperature strength of the hard alloy from that of the base material. This appeared to be a more logical explanation. He states that, since the hard alloy was quite strong and relatively brittle, it cracked rather than yielded when thermal stresses exceeded its strength.

One investigator<sup>(H55)</sup> described an unusual type of transverse weld-metal cracking in butt welds made with a modified SAE 4315 (Cr-Ni-Mo-V) steel electrode on SAE T 1330 (0.30% C, 1.90% Mn, 0.35% Si, Bal. Fe) steel plate material. These cracks were not visible immediately after welding, and X-ray inspection indicated that the welds were sound. The cracks began to appear about three to ten days after welding. The first visible indication was a narrow ripple (Luder line) or depression in the surface, at which point a crack later appeared. This investigator concluded that the transverse cracking started at the center of the weld, probably during welding, and was later propagated to the surface by residual stresses plus the added stresses built up by delayed transformation.

Another investigator encountered transverse cracking in the reinforcing beads of a weld deposited with stainless Type 25 Cr-20 Ni electrodes on steel armor plate. The cracks were purely transcrySTALLINE and

he believed that they were cold cracks. This investigator said that the cracking was directly caused by base-metal dilution, which formed an air-hardening alloy whose suppressed transformation temperature developed sufficiently high internal stresses to produce cold cracks.

### Multidirectional Cracks

Some types of cracks in weld metal propagate in all directions and do not seem to bear any particular relation to the direction of welding. An example of this type of cracking is the so-called "check cracks" encountered in austenitic stainless steel weld deposits<sup>(H45, I55)</sup>. These cracks are generally very fine and extend in all directions. They are visible on the surface only in extreme cases, as shown in Figure 22. Even the use

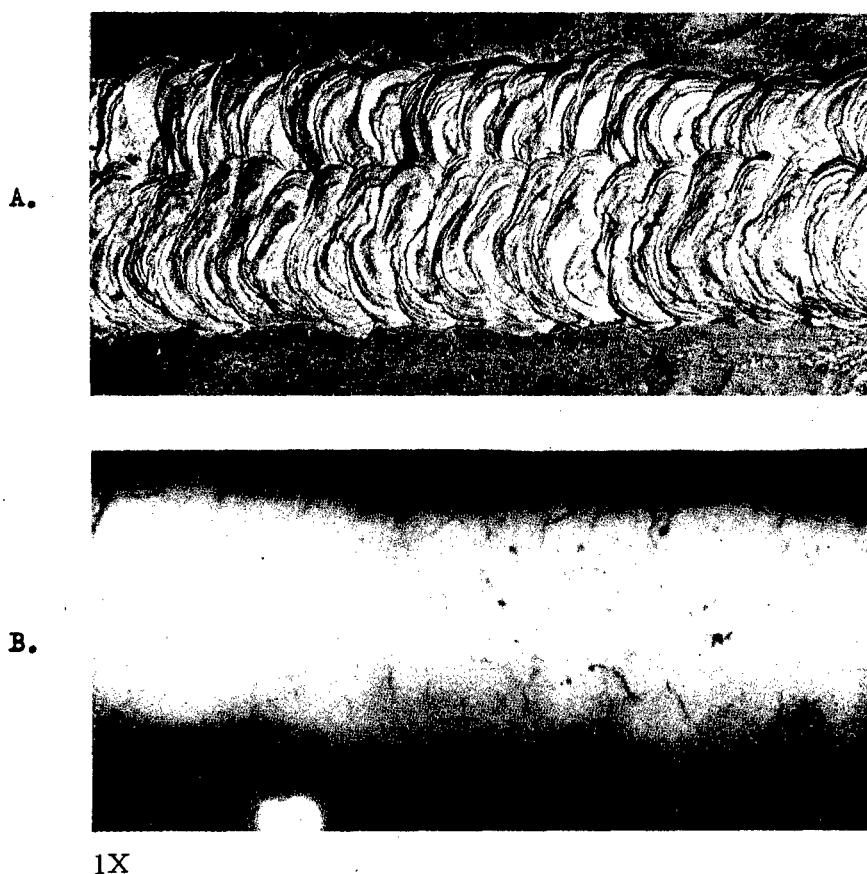


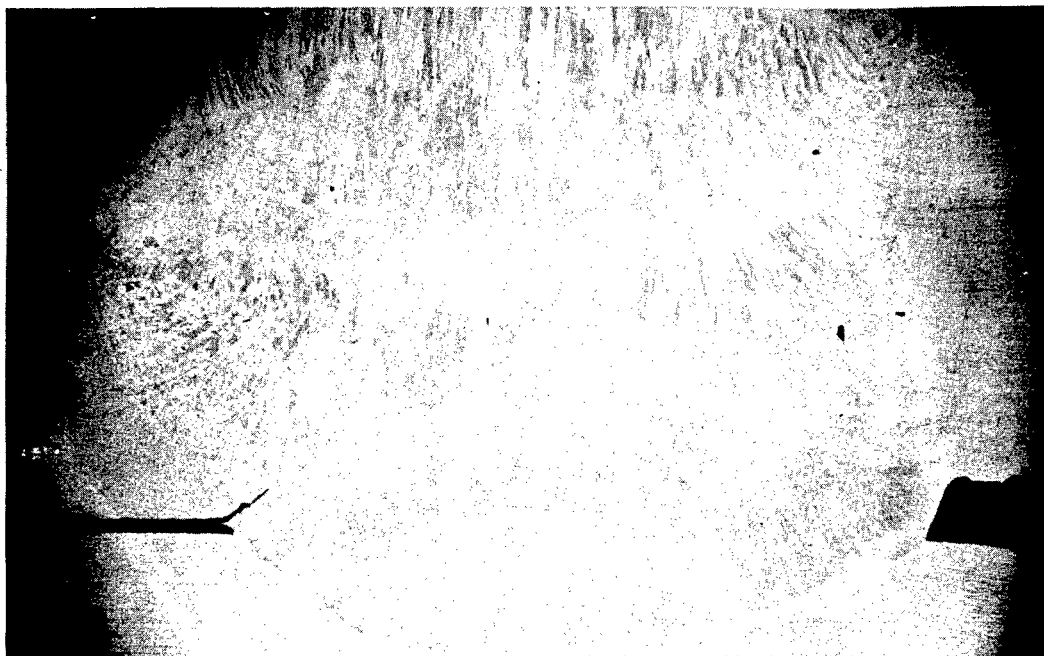
FIGURE 22. SURFACE APPEARANCE AND X-RAY PHOTOGRAPH OF WELD METAL CONTAINING MULTIDIRECTIONAL CHECK CRACKS<sup>(I45)</sup>

of X-rays or gamma rays may not reveal the extremely fine cracks. They probably result from extremely weak grain boundaries, probably in the hot-short temperature range.

## Notch Cracks

Notch cracks in weld metal occur at points of stress concentration such as those created by incomplete root fusion<sup>(H122)</sup>, interfaces between adjacent base plates, entrapped slag in the weld metal<sup>(H51)</sup>, and other factors<sup>(H261)</sup>. Notches are cracks and, as such, tend to propagate themselves. After cracks start to lengthen, the resultant reduced cross section and increased concentration at the ends of the crack usually make matters progressively worse<sup>(H288)</sup>.

One type of crack often encountered in welding is that caused by the notch at the root of a weld when using a backing strip<sup>(H58, H104)</sup>. The cracks are not confined to the root bead and may spread into later beads. One of these cracks is shown in Figure 23.

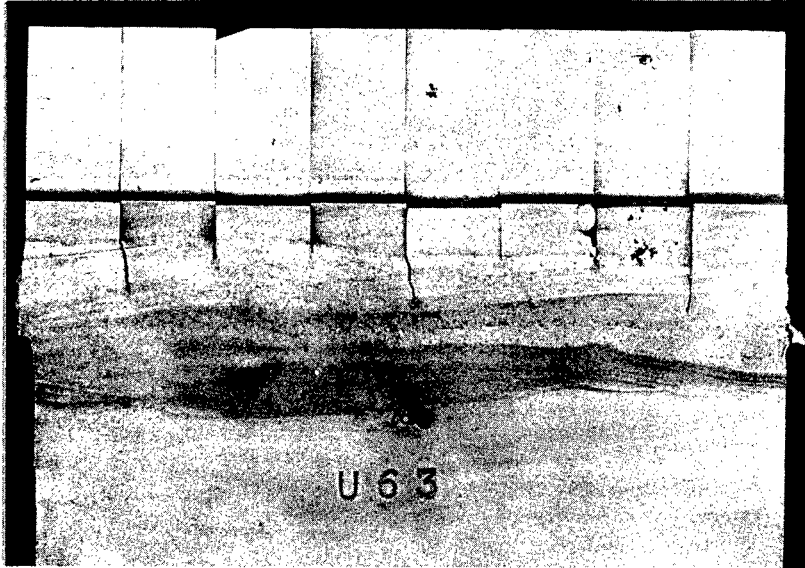


7- $\frac{1}{2}$ X

2% Nital Etch

FIGURE 23. SECTION OF WELD JOINT IN  $\frac{1}{2}$ -INCH SAE 4130 STEEL SHOWING ROOT CRACK AT JUNCTION OF WELD, BACKING BAR, AND BASE PLATE<sup>(H58)</sup>

Another good example of notch cracking is the interbucket extension cracking encountered during the metal-arc welding of buckets to gas-turbine wheels<sup>(I9, I21, I22, I51)</sup>. These cracks initiated in the weld metal as extensions of planes of contact between adjacent buckets and sometimes propagated across the welds and into the wheel section during service at elevated temperatures. A wheel-and-bucket replica specimen containing notch-extension cracks is shown in Figure 24.



1X

FIGURE 24. NOTCH EXTENSION CRACKS IN WELDED WHEEL- AND-BUCKET REPLICA SPECIMEN<sup>(151)</sup>

### Fissure

The term "fissure" has been used to describe several types of microscopic cracks or defects in weld metal which have a detrimental effect on strength and ductility. Fissures in austenitic weld metal do not appear to be due to the same causes as fissures in ferritic weld metal, so the two will be discussed separately.

In austenitic weld metal, fissures have been defined as random minute discontinuities that are believed to form during welding or shortly afterwards, while the weld metal is still very hot<sup>(H55)</sup>. They may be too small for X-ray detection but may show up later during testing of an all-weld-metal tensile specimen in the form of numerous small ruptures on the surface as shown in Figure 25. The microfissures are intergranular in nature<sup>(I34)</sup>. Attempts to show that fissures could be eliminated by quench-annealing heat treatments were unsuccessful, thus confirming the theory that the fissures form during the solidification of the weld metal<sup>(I34)</sup>.



FIGURE 25. FISSURES IN FRACTURED ALL-WELD-METAL TENSILE SPECIMENS OF TYPE 310 STEEL<sup>(15)</sup>

The exact cause of fissures in austenitic weld metal has not yet been clearly established. The results of an investigation by Carpenter and Jessen<sup>(15)</sup> indicated that silicate films, as shown in Figure 26, were chiefly responsible for fissuring and failure of 25 per cent Cr—20 per cent Ni weld metal.

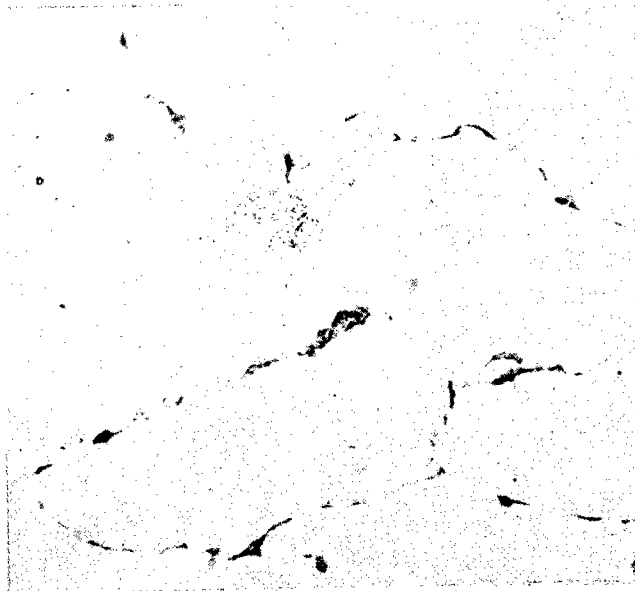


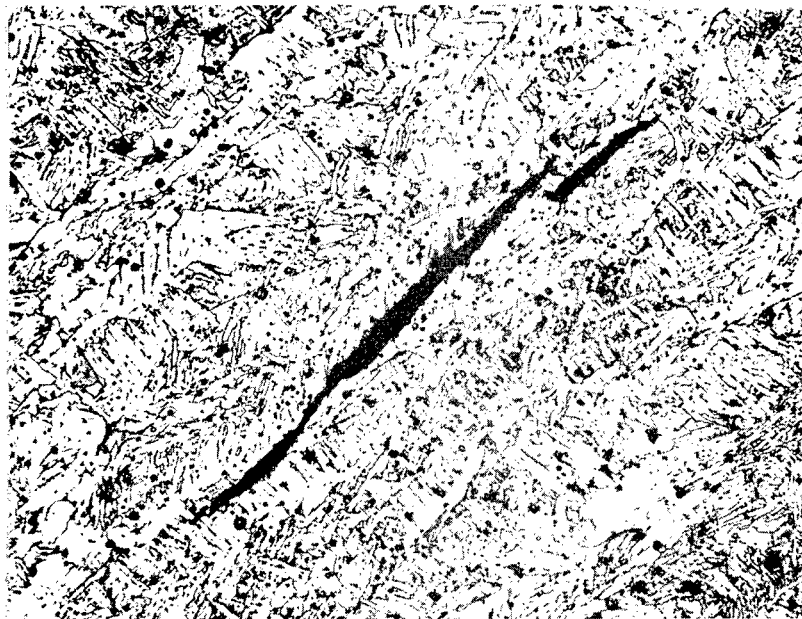
FIGURE 26. INTERGRANULAR SLAG FILMS IN HIGH-SILICON TYPE 310 WELD METAL<sup>(15)</sup>

Thomas<sup>(H279)</sup> stated that an austenitic weld metal with strong fissuring tendencies seems to be most subject to high-temperature cracking.

This suggests the possibility that the cracking may start at a microfissure while the weld metal is cooling to room temperature.

Fissures in ferritic weld metal reported in the literature appear to be caused by a phenomenon different from that causing fissures in austenitic weld metal. Miller and Kovac<sup>(H183)</sup> reported the occurrence of microfissures in multiple-bead arc welds made with low-carbon steel (0.17% C, 0.55% Mn, 0.07% Si, 0.022% P, 0.021% S) electrodes on fire-box plate steel (0.19% C, 0.50% Mn, 0.075% Si, 0.016% P, 0.026% S). The authors described the fissures as being intercrystalline, and often associated with a small particle of slag. The fissures appeared to be restricted to the fine-grained, recrystallized regions of the weld, none being found in the coarse-grained columnar weld metal. These fissures were believed to form at temperatures somewhat above the equicohesive temperature (the temperature at which the grain-boundary strength attains the same strength as the grain, on cooling).

Flanigan<sup>(H93)</sup> reported the occurrence of microcracks or fissures as part of an investigation on the effects of hydrogen on the ductility of weld metal deposited with AWS Class E6010 electrodes. These cracks were observed in single-bead deposits which were quenched immediately after welding. One of these fissures is shown in Figure 27. Flanigan



250X

FIGURE 27. MICROCRACK IN E6010 WELD DEPOSIT (SINGLE PASS) INDUCED BY QUENCHING AFTER WELDING<sup>(H93)</sup>

reported, "these fissures or cracks tended to follow grain-boundary and interdendritic paths". He concluded that these fissures were responsible for the low ductility of rapidly cooled welds. The fissuring and low ductility was directly related to the hydrogen content of the weld metal. In later

work, Flanigan<sup>(H95)</sup> stated that the mechanism by which hydrogen produced weld-metal embrittlement was not fully understood. He felt that other factors, such as retention of austenite, or the precipitation of nitrides or carbides, may also be involved.

Work by other investigators has further indicated that hydrogen is a factor in causing fissuring in ferritic weld metal. Bland<sup>(H37)</sup> showed that the rate of cooling, as controlled by quench time, of welds deposited with AWS Class E6011 electrodes was directly related to fissuring. He likened the grain-boundary fissures in all-weld-metal tensile fractures to hydrogen flakes in forged steels. In single bead welds, the fissures were located, in general, at the boundaries of the columnar dendrites. The recrystallization and annealing of the bead that occurred during multipass welding produced no visible changes in the location and appearance of the fissures. In multipass welds, the fissures were predominantly confined to the fine-grain area of the recrystallized zones, although some were occasionally located in the large grains.

A typical fissure in recrystallized weld metal is shown in Figure 28.

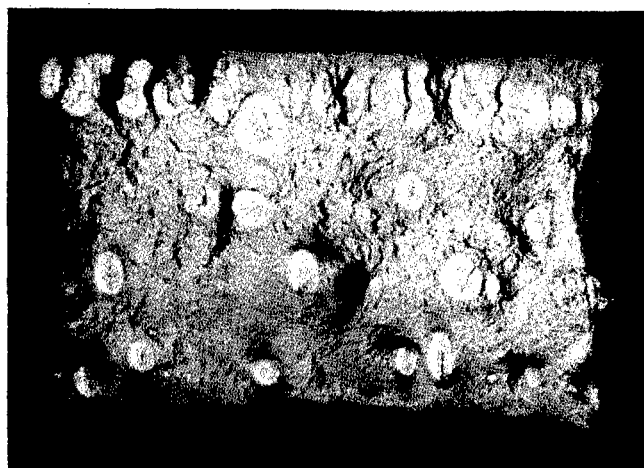


FIGURE 28. PHOTOMICROGRAPH OF FISSURE IN WELD METAL DEPOSITED WITH AWS CLASS E6011 ELECTRODES<sup>(H37)</sup>

Rollason and Roberts<sup>(H214)</sup> investigated the effect of cooling rate and composition on the embrittlement of weld metal deposited with low-carbon steel electrodes. Their results indicated that nitrogen, as well as hydrogen, in weld metal caused fissuring.

## Fisheyes

Fisheyes are small cavities which sometimes appear singly or in clusters in the fractures of all-weld-metal tension specimens. Zapffe<sup>(H296)</sup> described them as local embrittled areas surrounding small blowholes or inclusions. A photograph of a weld-metal fracture containing fisheyes is shown in Figure 29. A fisheye, if stressed during welding, might produce a fissure.



3X

FIGURE 29. FISHEYES IN FRACTURED WELD METAL<sup>(H297)</sup>

The occurrence of fisheyes is accompanied by low ductility in all-weld-metal tension specimens<sup>(H130)</sup>. However, both Ronay<sup>(H216)</sup> and Hopkins<sup>(H130)</sup> concluded that "fisheyes" were not defects, since they were seldom observed in fractures of specimens which had been thermally stress relieved. They were also avoided by using an interpass temperature of 300 F.

The small blowholes in the centers of the fisheyes are shiny on the interior, which suggests the presence of a reducing gas, such as hydrogen. The embrittled zone surrounding the small blowhole is believed to be caused by the presence of hydrogen which causes a localized stress condition<sup>(H297)</sup>. During straining, the localized stress area probably fractures slightly in advance of the remainder of the weld metal, due to the stress and the fact that the metal is embrittled.



## FACTORS WHICH INFLUENCE WELD-METAL CRACKING

Basically, weld-metal cracking occurs only when stress exceeds strength. It is reasonable to assume that, if a weld metal is able to flow plastically during cooling and thus remove stress concentration, its strength in some local area would not be exceeded by stresses generated in welding. Many weld cracks are brittle failures, which show little or no signs of ductility in spite of a very low flow strength at elevated temperatures. Hence it follows that cracking depends on some combination of the stress pattern and the characteristics of the weld metal which inhibit deformation. The greater the restraint, the greater must be the plasticity of the weld metal if cracking is to be avoided, or the freer it must be of structural and compositional characteristics which make it vulnerable to low stresses.

Stress in a weld during fabrication is mainly a function of joint design and temperature gradients. The metallurgical properties of the weld metal control its strength.

### Stress

Stresses in a welded joint originate primarily from forces of expansion and contraction, though external forces may also be superimposed. In some cases, shrinkage stresses may be partly relieved by plastic flow, as shown by the changes in size and shape produced during welding, and also warping and distortion. However, under conditions of severe external restraint and improper joint design, the stresses may locally approach the strength of the weld metal, even with high-quality metal, at a certain temperature. A rough idea of the amount of stress which can be generated can be had from considering the total amount of shrinkage during cooling from the molten state. This is slightly above one per cent (H223), or enough at room temperature to exceed the 0.1 to 0.2 per cent limit of elastic elongation, but much below the 15 to 40 per cent elongation at failure obtained in most all-weld-metal tensile specimens. Nevertheless, cracking failures do occur in weld metal at temperatures approaching room temperature. Apparently, through some mechanism, the normal plastic yield of the weld metal is inhibited.

A condition of multiaxial shrinkage stresses has been used by many investigators to explain the inability of weld metal to exhibit plastic flow in some welded joints (H16, H210, H223, H235, H260, H288). If shrinkage in a weld were under rigid multiaxial (complete triaxial, with all three stresses equal) restraint, it is obvious that the metal would either crack or leave some kind of a shrinkage cavity. Because of this, ductile steel may be made to behave in a brittle manner depending upon the way in which it is stressed (H288).

In the welding of relatively thick and rigid plates, triaxial stresses arise due to shrinkage laterally, longitudinally, and vertically<sup>(H210)</sup>. Plastic flow is normally produced by shearing stress but, in the presence of triaxial tension of equal magnitude in each direction, shearing stresses are absent, plastic flow is prevented, and a brittle fracture occurs as soon as the cohesive strength of the material is exceeded.

Hartog and Rodgers<sup>(H111)</sup> used a rather novel method to measure the internal stress distribution in a welded joint in an attempt to determine the causes of cracks in welded structures of ductile boiler plate. Weld metal was deposited on a strip of steel about one quarter of an inch thick which had been clamped over the plates to be treated. The arc did not penetrate to the plates below, but did raise the temperature high enough to cause the metal to reach its yield point in compression and produce internal tension upon cooling to room temperature. The metallurgical structure of the plates was not affected by the weld deposit. After the welding operation, the specimens were cut up in segments to determine the stress distribution by measuring the relaxation of strain. The stress pattern consisted of a transverse tensile stress in the heated zone almost equal to the yield point, which was balanced by areas of low compressive stress on both sides of the heated zone. As far as the problem of weld-metal cracking is concerned, these results served to illustrate the high tension stress that develops at the centerline of a weld.

Nippes and Savage<sup>(H188)</sup> measured the transverse stress on the surfaces of welds which were held in restraint almost to the point of failure by cracking. Their measurements were made after the welds cooled to room temperature by applying SR-4 resistance strain gages to the base plates and then observing the relaxation of strain after the welds were cut. The welds were made in  $\frac{1}{4}$ - and  $\frac{1}{2}$ -inch-thick steel plates, using AWS Class E6020 or E6010 electrodes. These investigators stated that transverse stresses approached 65,000 psi, the ultimate strength of the deposited weld metal. Longitudinal stresses of approximately 40,000 psi were obtained which equalled the yield strength of the weld metal.

### Joint Design

Some welding authorities have stated that, if welded joints were perfectly designed as far as stress concentrations were concerned, there would be no reason to worry about the other factors<sup>(H288)</sup>. Much of the confusion in design for welding is due to the fact that each joint is a problem in itself and that all variables, such as thickness of plate, speed of travel, rate of power input, and other factors have their individual effects<sup>(H141)</sup>. Many of the early welding troubles originated from faulty design which was frequently based on experience with riveted structures<sup>(H210)</sup>. Restraint of weld-metal contraction is probably the greatest

single factor in causing welds to crack and pull apart during fabrication. Even the most crack-resistant weld metal can stand only a small amount of stress when cooling through the liquidus-solidus range, and, if this critical stress is exceeded, the weld metal will crack.

The total amount of restraint in a welded joint is controlled mainly by the joint geometry, the thickness of plate, and the sequence of welding (H122). Heat input is also a secondary factor, as discussed later on in this report.

Circular joints are particularly difficult to weld without longitudinal weld-metal cracking because of the high restraint imposed. Examples of such joints are: circular patches on boilers<sup>(H123)</sup>, shafts through plates and wheels<sup>(H52)</sup>, and straps on heavy-wall pipe<sup>(H273)</sup>. Transverse stresses of 40,000 to 50,000 psi have been measured in such a joint<sup>(H123)</sup>. Another troublesome joint, often used in the aircraft industry, is the attachment of welding lugs or fins longitudinally on heavy-wall tubes or shafts.

Root Gap. A quality job is impossible without good fitup. The proper adjustment of the root gap is important in preventing weld-metal cracking. A wide root gap has the effect of increasing the over-all contraction of weld metal and hence of causing the tendency to crack at the center of the weld, which is last to solidify<sup>(I2)</sup>. A narrow root gap is also bad since tightness often leads to cracking<sup>(H55, H243)</sup>. In a tee-joint, for example, rough flame-cut surfaces will cause less trouble than finished machined surfaces<sup>(H123)</sup>. A narrow root gap actually causes notch-extension cracking, as discussed below.

Scarf Angle. The scarf angle used in the preparation of base plates for welding is important in controlling weld-metal cracking. Too narrow a groove may introduce a notch at the root and also lead to incomplete penetration and slag entrapment which invite cracking<sup>(H51)</sup>. Large scarf angles are bad because of the greater dilution effect<sup>(H55, H260)</sup>.

Land. A wide land or lip preparation on base plates reduces dilution of the base metal by the weld metal and therefore indirectly reduces weld-metal cracking. Cracking in welds made at Battelle on armor using 19-9 Mn stainless steel electrodes appeared to be reduced by increasing the width of the land from 1/16 to 1/4 inch.

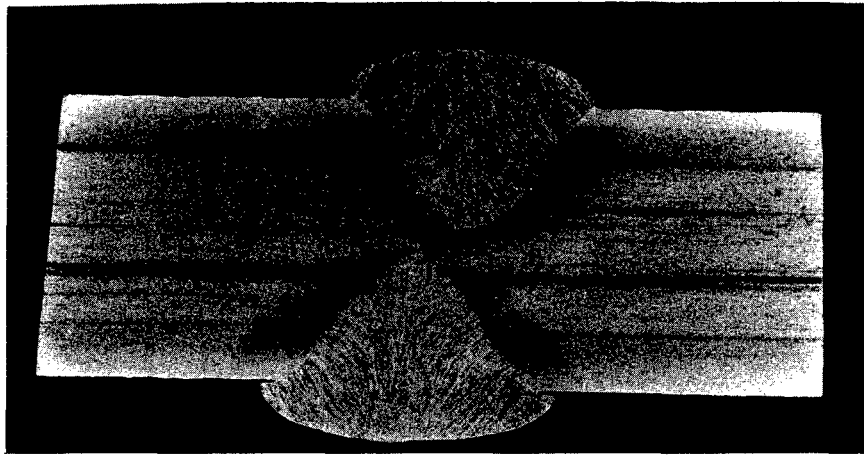
Backing Strip. Backup bars of various kinds are needed for certain types of weld joints. The design of backups is very important since they may have various effects on weld-metal cracking. One is the possible

notch effect caused by the interface between the base plate and the backup and another is the chilling effect. The latter is dependent upon the mass of the backup and the material of which it is made.

Root cracking was encountered in single-vee butt-joint welds made with AWS Class E10020 electrodes on SAE 4130 plate<sup>(H58)</sup>. An example of one of these cracks was shown in Figure 23. These cracks were essentially a result of the discontinuity produced by the backing strip. Similarly, another instance of root cracking was attributed to improper fitup of integral chills in power-plant piping<sup>(H55)</sup>. The cracking was prevented by using machined, close-fitting chills that were held tightly in place. These chills were vee-shaped to conform with the bottom half of a double-vee groove.

Plate Thickness. Thick plates are more difficult to weld, without encountering weld-metal cracking, for two reasons. First, the cooling of the weld metal is more rapid than in welds deposited on the plate because of the greater mass, and second, the rigidity of the thicker plates causes a greater amount of restraint on the weld metal, which leads to cracking.

Notches. Notches increase the likelihood or incidence of weld-metal cracking since they act as stress raisers<sup>(H117)</sup>. Notches arise in welds in the form of inclusions, undercut, overlap, and interfaces between two base plates. Notches create a condition of multiaxial stress which may result in brittle failures at the notch. The stress intensity at the end of a small crack, for example, may be many times the general stress intensity in other parts of the weld. As discussed in the section on types of cracks, a particularly severe type of notch crack is the notch-extension crack encountered in welding buckets to high-temperature gas turbines. An example of this type of cracking was shown in Figure 24. Laminations in base metal sometimes start cracks in the weld metal both because of the notch and because the lamination is a point of segregation<sup>(H52)</sup>. One of these cracks is shown in Figure 30.



2X

HCl Etch

FIGURE 30. CRACK IN FERRITIC WELD-METAL DEPOSIT STARTING AT A LAMINATION IN THE BASE METAL (H52)

Size and Shape of Weld Bead. The size and shape of the weld-bead deposit has considerable influence on weld-metal cracking. Thin root beads and concave fillet beads are particularly susceptible to cracking because of their weak cross section (H15, H122, I45). Convex fillet beads were found to be least susceptible to longitudinal cracking (H188, H271, H272). Stout, et al (H271), concluded that a high narrow bead resisted cracking better than a thin, wide bead. However, other investigators (H116, H260) found that convex beads often contain root cracks which do not extend to the surface and thus escape detection. Another investigator (I17) recommended using heavy bead deposits to minimize cracking. Somewhat contrary to this, one investigator (H262) stated that cracks in welded structures were often due to heaping up of weld metal at complicated joints. A flat fillet-bead deposit was generally believed to be the most reliable as far as weld-metal cracking is concerned (H116).

Penetration. Incomplete penetration in some weld joints may lead to weld-metal cracking (H15). It is caused by the failure of the root face of a groove weld to reach the fusion temperature for its entire depth. The condition is a result of poor welding technique, such as using a large electrode in a narrow root opening. Improper heat distribution is another factor,

since areas of base metal above the root may reach the melting temperature before the proper temperature has been attained at the root.

One investigator<sup>(H88)</sup> reported difficulties with weld-metal cracking due to incomplete penetration. He stated that when penetration in a weld was only half complete, about four times the normal stress was transmitted to the weld. The difficulties due to incomplete penetration can often be corrected by beveling the base plates at the joint.

### Heat Input and Cooling Rate

The effect of total heat input on weld-metal cracking is not entirely clear. In some instances, a high heat input increases cracking, but in other instances, it does not, depending somewhat on the weld-metal composition and its relative susceptibility to cracking. Generally, it may be said that excessive heating of the molten weld metal puddle will increase the possibility of cracking. Rollason<sup>(H215)</sup> stated that superheating of the weld results in larger grains and less grain-boundary area to accommodate precipitated sulfides.

It is generally known that weld metal deposited with the so-called "hot" AWS Class E6020 and E6030 electrodes are more susceptible to longitudinal root-bead cracking than the corresponding E6010 and E6012 electrodes. This is thought to be due to the higher currents used with the "hot" electrodes and the resultant slow rate of solidification. On the other hand, differences in the metallurgical properties of the weld metal, as discussed later in this report, may be a more important factor than the total heat input. Spraragen<sup>(H260)</sup> cited several instances of "hot" cracking which appeared to be caused by high heat inputs during welding. This may be an indirect effect, since it is known that the "hotter" electrodes deposit thinner root and fillet beads, which are more susceptible to cracking<sup>(H272)</sup>.

Butterfield<sup>(I2)</sup> found that superheating of the molten weld pool increased hot cracking in weld metal deposited with austenitic electrodes (18 Cr-8 Ni). He stated that the cracking might be due to the slow cooling rate through a certain "critical" temperature range which allowed greater opportunity for the application of stress while the weld metal was in a weak state.

Generally, it was believed that slow cooling rates, as effected by high heat input, should lessen the chances of "hot" cracking because lower internal stresses would be set up during cooling. However, in tests with Lehigh Restraint specimens, it was found that heat input in welds deposited with AWS E6010 and E6020 electrodes seemed to have no detectable relation to the stress developed or to the temperature of cracking (below 393° F)<sup>(H271)</sup>. Cold cracking in ferritic weld metal occurred only when the cooling rate was increased by quenching<sup>(H37, H96, H215, H264)</sup>.

Preheat and Postheat. Preheat is used quite often in welding because it reduces the temperature gradient between the weld and the base metal and thus lowers the stress that leads to weld-metal cracking. Preheat also lowers the cooling rate after welding. Immediate postheating is also effective in decreasing the cooling rate, particularly at lower temperatures. It is used mostly as a stress-relief treatment to reduce distortion and the formation of hard zones and shrinkage stresses. There is evidence that preheating and postheating allow hydrogen to diffuse out of weld metal and also effect transformation of austenite at a higher temperature (H96).

Rollason (H215) felt that preheating did not always improve the resistance of weld metal to cracking. His opinion was that preheating increased hot cracking in mild-steel weld metal by causing the manganese to remove sulfur from solution, and thus increasing the formation of low-melting films of (FeMn)S. However, he stated that, with high-rutile electrodes Type E217 (similar to AWS E6012), preheating might not have a detrimental effect because the  $2\text{MnOSiO}_2$  inclusions would have a higher melting point and would not be likely to cause cracking.

Preheating for austenitic weld metals (both in austenitic and ferritic base plates) is usually not necessary. However, preheating has been used to eliminate cracking during the welding of heavy sections of columbium-bearing Type 347 steel (I40).

Preheating and postheating have definitely been shown to be useful in preventing cold cracking in weld metal, as they are in preventing under-bead cracking in the base metal. For example, Steinberger, et al (H264) found that cold cracking did not occur in weld metal deposited with SAE 4330, 4320, 6130, 4130, and N.E. 8630 filler rods, if the welds were postheated for one hour at 400°F or above and then water cooled to room temperature. Cracking occurred in all the ferritic welds given similar postheat treatment at temperatures lower than 400°F. Flanigan (H96) found that short-time postheating temperatures above 850°F increased the ductility of AWS Class E6010 weld deposits, by eliminating hydrogen. Treatments at 250° to 650°F also increased ductility, but did not entirely eliminate hydrogen. Rollason (H216) recommended preheat temperatures of 212°F to eliminate fissuring due to hydrogen in mild-steel weld metal deposited on thick plates.

Welding Speed and Current. Careful adjustment of the welding speed and current are important in metal-arc welding for control of weld-metal cracking. If the welding speed is too high for the normal deposition rate, the cross section of the deposited bead will be too small and weak to accommodate the cooling stresses. Nippes (H123) attributed cracking in highly restrained weld metal deposited with the atomic-hydrogen welding process, using SAE 6330 filler wire on SAE 4320 steel propellers, to increased speed of welding. Helin and Svantesson (H116) believed that hot

cracking increased with the welding speed because it favored poor root penetration and entrapment of slag in the weld metal.

It is generally known that high welding currents increase the hot-cracking tendency of weld metal. This may be due to superheating of the molten weld metal<sup>(H216)</sup>, though Stout, et al<sup>(H271)</sup>, attributed its effect to increased base-metal dilution.

Butterfield<sup>(12)</sup> determined the effect of welding current on a large number of electrodes and found that hot cracking definitely increased with the welding current (no mention was made of testing technique or data). Thomas<sup>(145)</sup> noted that fissures in austenitic weld metal deposits were more numerous and of greater size when high welding currents were used. Carpenter and Jessen<sup>(15)</sup> concluded that increased fissuring in austenitic (25 Cr-20 Ni) weld metal was caused by oxidation of silicon at excessively high welding currents.

Electrode Size. Large electrodes cause cracking in austenitic weld-metal deposits because the large pools of molten metal deposited cause excessive shrinkage, slower cooling, and a large grain size<sup>(H254)</sup>. Butterfield<sup>(12)</sup> stated that, if two sizes of 18 Cr-8 Ni electrodes deposit the same size fillet, the larger electrode will show the greater tendency to hot cracking, but if the fillet size is maintained proportional to the electrode size, there appears to be little difference. This may be an indirect effect, since the mild weaving necessary with the smaller size electrode tends to prevent cracking.

Electrode Coating. The slag covering provided by the coating on metal-arc electrodes tends to slow down the cooling of the weld-metal deposit and thus increases the effect of the total heat input. One investigator<sup>(H260)</sup> stated that increased cracking in weld metal deposited with coated electrodes, over that deposited with bare electrodes, was due to the added heat input. Several investigators<sup>(12, H116)</sup> concluded that excessive slag from the coating caused concave weld-bead deposits which were susceptible to cracking. Claussen<sup>(H60)</sup> found that fillet welds made with the heaviest coated electrodes were concave and had severe hot cracks through the throat immediately after welding.

Coatings on AWS class ferritic electrodes have various effects on the tendency to weld-metal cracking. Weld deposits made with EXX20 and EXX30 electrodes are more subject to cracking than any of the other classes, and therefore, these electrodes should not be used in depositing the root pass<sup>(H51)</sup>. In double-vee groove joints, it is common practice to use EXX10 electrodes in the first pass, and then fill in the subsequent passes with EXX20. The EXX20 and EXX30 electrodes tend to deposit concave beads which are weak and likely to crack. The weld deposits are



also more susceptible to sulfur cracking, as discussed later in this report. The high susceptibility of the EXX20 weld deposits to cracking was very effectively illustrated in Figure 16.

AWS Class EXX10 weld deposits are relatively free of cracking troubles if sulfur segregation is not encountered.

Welds deposited with AWS Class EXX15 and EXX16 electrodes have the greatest resistance to hot cracking which is due to impurities such as sulfur. However, the coatings on these electrodes form a viscous slag which is conducive to slag entrapment<sup>(H51)</sup>. These electrodes should, therefore, not be used in narrow groove joints.

Very little information on the relative cracking susceptibility of the other classes of electrodes was found in the literature.

Plate Thickness. In addition to its effect on restraint, the thickness of the base metal in a welded joint has considerable effect on the rate of cooling of the weld metal. Generally, the cooling effect increases with plate thickness. Hydrogen embrittlement in mild-steel weld deposits was attributed to the increased rate of cooling caused by thick plates. In an investigation to determine the cooling rate in the heat-affected zone of thick plates, it was found that a 1-inch-thick plate behaved as an infinitely thick plate<sup>(H121)</sup>. A 1-1/2-inch plate cools no more rapidly than a 1-inch plate.

## Metallurgical

What needs to be learned are the fundamental metallurgical factors associated with weld-metal cracking, and the reasons why one type of weld metal is more susceptible to cracking than another. Since most weld-metal cracking is held to be intergranular or interdendritic, we might expect to find a brittle or low-melting constituent in the grain boundaries of crack-sensitive weld metal. The properties of this constituent would be largely controlled by composition. A secondary effect might be grain size, since the distance between the grains increases with grain size.

In general, austenitic weld metal shows a greater susceptibility to hot cracking than ferritic weld metal. Castings of austenitic steels are known to hot tear more readily than cast ferritic steels. The reasons for this have not been satisfactorily explained. It is believed, however, that the high-alloyed austenitic weld metal is more subject to segregation<sup>(I23)</sup>.

Certain alloying elements do not exert the same effect on ferritic weld metal as they do on austenitic weld metal. Therefore, to avoid confusion, it is necessary to discuss the effects of composition on the two types of weld metal separately.

### Ferritic Weld Metal

Various alloying elements are added to ferritic weld metal to enhance its properties. Other elements get into the weld deposit as residual impurities which could not be avoided in commercial steelmaking practices. The following discussion is concerned with the effects of these alloying elements and impurities on the susceptibility to weld-metal cracking.

Carbon. Carbon reduces the high-temperature ductility of steel and, therefore, would be expected to influence hot cracking in weld metal. Helin and Svantesson<sup>(H116)</sup> believed that carbon increased the cracking susceptibility because of its high solubility in liquid iron and its marked tendency to segregate. Carbon has a great effect in fully killed weld metal, which partially explains the detrimental effect of silicon deoxidation on weld-metal cracking. It was shown that hot cracking in ferritic weld deposits increased with the carbon content of the base metal (dilution into the weld metal)<sup>(H116)</sup>. Helin and Svantesson<sup>(H116)</sup> further suggested that the effect of carbon might be attributable to internal pressure. They stated that the theoretical internal pressure produced by the formation of CO is a direct function of the amount of carbon and FeO present in the weld metal. Helin and Svantesson<sup>(H116)</sup> suspected that the CO pressure developed too high a value during final solidification of the weld metal.

Cold cracks in alloy steel weld metal deposited with the atomic-hydrogen welding process appeared to be directly influenced by the effective carbon content as determined by circular-groove cracking tests<sup>(H264)</sup>. It is generally believed that carbon increases the sensitivity of ferritic weld metal to hydrogen cracking. In these tests, Type 6130 (0.30%C, 0.80%Mn, 0.35%Si, 0.95%Cr, 0.22%V) weld metal showed less cracking than Type 4330 (0.30%C, 0.70%Mn, 0.30%Si, 1.80%Ni, 0.70%Cr, 0.32%Mo) weld metal, although their carbon contents were the same. The investigators concluded that the vanadium in the Type 6130 weld metal promoted the formation of stable carbides which did not readily go into solution during the rapid thermal cycles involved during welding. Tying up a portion of the carbon content of the weld metal makes it respond in a manner similar to that of a material of lower carbon content. In discussing this investigation, Bland<sup>(H37)</sup> disagreed somewhat with the authors' conclusion. He cited as an example, the high crack sensitivity results with Type 410 (0.15%C, 12.50%Cr, bal. Fe) weld metal which should have had all of the carbon tied up as stable carbides. In another discussion, Sims<sup>(H238)</sup> pointed out that if austenite is retained which does not transform at room temperature (as is the case in austenitic stainless steel), it acts as a reservoir to hold hydrogen in solid solution where it does no harm.

Silicon. Silicon increases the tendency toward weld-metal hot cracking, chiefly owing to its deoxidizing action, according to Reeve<sup>(H201)</sup>. This investigator also suggested that silicon acts in a secondary manner by promoting the segregation of sulfur.

Rollason<sup>(H215)</sup> offered some evidence, based on tests with the Murex hot-cracking test (discussed later in this survey), that the cracking tendency of ferritic carbon-steel deposits increases as the silicon increases. His results are shown in the following tabulation:

Specimen No.	Composition, per cent					Crack Length, inches
	C	Mn	S	P	Si	
Y	0.11	0.50	0.026	0.056	0.135	0
R	0.10	0.53	0.128	0.042	0.162	2

Note, however, that the 0.10 per cent difference in sulfur content was certainly much more important than the 0.027 per cent difference in silicon content.

Manganese. Manganese is beneficial in reducing the hot-cracking tendency of ferritic weld metal. The function of manganese is generally

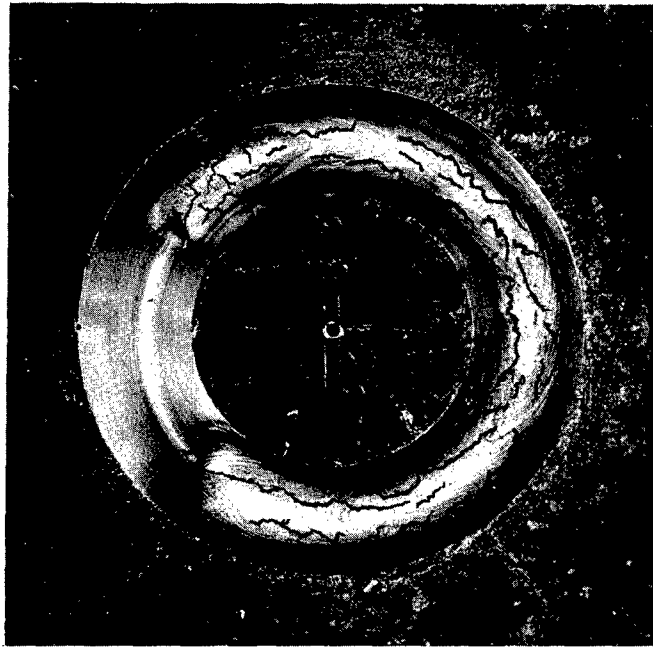
believed to be in forming MnS, which has a higher melting point (2915°F) and a much lower solubility in liquid iron than FeS(H60). To convert all of the sulfur in weld metal to MnS, at least 1.75 parts of manganese should be added for every part of sulfur. However, a mixed (FeMnS) sulfide containing 25 per cent MnS can cause hot cracking because of its low melting point (H215). To be on the safe side, most authorities recommend a minimum ratio of 5 parts Mn to 1 part sulfur. In most cases, even more manganese is used. Helin and Svantesson(H116) report that a weld containing 1.02 per cent manganese and 0.47 per cent silicon was resistant to hot cracking when deposited on parent metal containing up to 0.1 per cent sulfur. The electrode used to produce such weld metal had a highly basic slag, however, which appeared to favor resistance to sulfur hot cracking. Reeve(H202) did not note any appreciable improvement in the crack resistance of 0.10 per cent carbon-steel weld metal when manganese was added. He believed that any beneficial effect of the manganese was counteracted by a corresponding increase in silicon content. Silicon normally deoxidizes weld metal and causes greater sensitivity to checking due to sulfur.

Tests with fillet-welded single-tee-joint cracking specimens showed that manganese prevented cracking in welds deposited with AWS Class E6015 electrodes, even with 0.086 per cent sulfur in the weld pad(H60). This was supporting evidence that weld metal deposited with low-hydrogen lime-ferritic type electrodes was less susceptible to hot cracking due to sulfur. It may be that the basic slag tends to take up sulfur.

Cracking in weld metal deposited by the submerged-arc welding process was attributed to low-melting eutectic compounds of MnS-MnO-FeO-FeS(H136). The cracking was eliminated by increasing the manganese content (see Figure 31).

Sulfur. Sulfur (and its compounds) as an impurity is believed to be a major factor in causing weld-metal cracking. The effects of many other elements are often attributed to their influence on sulfur segregation in the weld metal. Since sulfur has a detrimental effect on both hot shortness and hot tearing in steel, it would be expected to be a strong factor in weld-metal cracking. One investigator(H59) believed that sulfur was not always the cause of trouble, but that its very presence showed that some other impurity might be there also.

It was suggested that the effects of sulfur are related to the formation of low-melting compounds such as FeS, which melts at 1805°F(H6). It is believed that interdendritic segregation of low-melting constituents which occurs during solidification of the weld metal, creates a zone of weakness which may lead to cracking. The low-melting sulfur-rich melt also tends to increase segregation of other impurities to the grain boundaries, since it remains molten for a longer period of time than normal during solidification of the weld metal(H116). Sulfur is known to increase the hot-tearing



A. Weld-Metal Composition 0.17% C, 0.48% Mn, 0.37% Si



B. Weld-Metal Composition 0.25% C, 2.56% Mn, 0.05% Si

FIGURE 31. CIRCULAR-GROOVE WELD-METAL DEPOSITS MADE WITH THE SUBMERGED-ARC WELDING PROCESS SHOWING THE EFFECT OF MANGANESE IN REDUCING CRACKING(H136). (Cracks were filled in with India ink to show them more clearly.)

tendency of steel. Sulfur also exerts a secondary influence because it increases the fluidity of the weld metal; this favors the formation of concave weld-metal beads as fillets, which are less able to withstand high shrinkage stresses than full or convex deposits(H116).

The Welding Handbook(H15) states that sulfur is usually kept below 0.05 per cent in steel, and causes little difficulty with welding because enough manganese is added to counteract its ill effects. It further states, however, that aggravated segregation, as with shielded-arc or submerged-arc welding, can lead to weld-metal cracking because of the melting out of complex sulfide inclusions from the base metal. It is not so much the amount of sulfur in the steel that is important for welding, as the form in which the sulfur appears(H51, H253). Even though the nominal sulfur content of the weld metal is low, there may be segregated areas of high sulfur content where failure by cracking could occur. One investigator(H51) made the assumption that manganese sulfide might be a cause of cracking because it combines with FeO, MnO, and FeS to form segregated patches of a low-melting eutectic which melts at about 1670°F. Another hypothesis(H239) was advanced that grain-boundary segregates of the eutectic compound Fe - FeS - Fe<sub>3</sub>C (mp 1780°F) caused both hot cracking and low mechanical-test results in ferritic weld metal. This eutectic was a suspected cause of brittleness in the temperature range 1650°- 1830°F.

Stieler(H266) made a series of fillet-welded tee-joint tests using bare and coated electrodes, both dusted with sulfur to increase the sulfur content of the weld deposit. The welds made with the bare electrodes were crack free, whereas the welds deposited with coated electrodes cracked, with the amount of cracking increasing with sulfur content. Chemical analysis of weld metal which contained cracks, showed 0.40 per cent sulfur in the bare-wire-electrode deposits and 0.15 per cent sulfur in the heavily coated electrode deposits. This indicated that bare-wire deposits could tolerate a higher percentage of sulfur than coated electrodes. Linetskii(H170) made similar tests in which sulfur was introduced into the weld metal by coating bare electrodes with sodium sulfate. Cracks appeared in the mild-steel weld metal when the sulfur content of the deposit exceeded 0.03 - 0.035 per cent and the ratio of manganese to sulfur was less than 5. Other investigators(H201) report that fillet weld metal in heavy plate tends to crack if the sulfur content exceeds 0.035 per cent. Cracking was unavoidable in weld deposits made on pressure vessels when the sulfur content was greater than 0.040 per cent, but no cracking was evident when the sulfur was kept below 0.025 per cent (no details)(H2).

Reeve(H201) made an extensive investigation into the effects of sulfur on weld-metal cracking. Using the fillet-welded tee-joint test, he determined the smallest amount of sulfur in a weld deposit which will produce cracking, for a series of welds on various carbon steels. He increased the sulfur content of each electrode type until cracking was encountered.

WADC TR 52-143 (CRACKING(H136)) (Cracks were filled in with Indis ink to show them more clearly.)

The results of these tests are shown in Table 6. Reeve concluded that the sulfur limit decreased with an increase in the carbon and other alloy content of the base metal.

An increase in manganese and especially silicon increased the crack sensitivity of the weld metal. Reeve pointed out that the effect of the manganese and silicon may have been in increased deoxidation.

TABLE 6. SULFUR LIMIT OF VARIOUS WELDING ELECTRODES AS INFLUENCED BY THE BASE-METAL CARBON CONTENT (H201)\*

Electrode	Normal Composition Weld Metal, per cent					Carbon in Base Metal, per cent	Sulfur Limit, per cent
	C	P	Mn	S	Si		
A	0.06	0.040	0.25	0.035	Trace	0.15	0.13
	0.06	0.040	0.25	0.035	Trace	0.24	0.09
	0.06	0.040	0.25	0.035	Trace	0.33	0.08
B	0.07	0.025	0.33	0.034	0.028	0.15	0.12
C	0.07	0.036	0.45	0.035	0.05	0.15	0.10
	0.07	0.036	0.45	0.035	0.05	0.24	0.06
	0.07	0.036	0.45	0.035	0.05	0.33	0.053
D	0.07	0.039	0.49	0.025	0.09	0.15	0.062
	0.07	0.039	0.49	0.025	0.09	0.24	0.06-0.069
	0.07	0.039	0.49	0.025	0.09	0.33	0.059
E	0.05	0.044	0.42	0.034	0.115	0.24	0.042
F	0.05	0.026	0.33	0.030	0.15	0.15	0.043
	0.05	0.026	0.33	0.030	0.15	0.24	0.042
	0.05	0.026	0.33	0.030	0.15	0.33	0.044
G	0.06	0.030	0.39	0.030	0.179	0.20	Probably below 0.05

\*Sulfur limit is defined as the smallest amount of sulfur in a weld-metal deposit which will produce cracking under the test conditions. In these tests, the sulfur content of each electrode was increased until cracking was encountered.

In the same work, Reeve also made metallographic studies of cracked and uncracked weld-metal deposits and made the following interesting observations:

1. The cracks were intergranular with some signs of oxidation, suggesting hot cracking.
2. The welds contained many small inclusions: silica, oxides, and mixed manganese-iron sulfides, which were higher in iron-sulfide in the cracked welds. Minute inclusions were more numerous in the cracked welds.
3. Large globular inclusions in a cracked weld were duplex in nature, consisting of oxides and silicates of varying silica content. In uncracked welds, the larger globules were monophased, chiefly oxides containing small and varying amounts of silica.
4. The larger globules were more numerous in the uncracked weld and appeared to have sulfide in solution.
5. Particles revealed on sulfur prints of cracked welds were more densely distributed and showed a more definite network tendency.

Sulfur prints of metal-arc welded joints are generally much paler in the weld-metal zone than in the adjacent base metal. The reason for this was believed to be the differences in the size of sulfide inclusions<sup>(H16)</sup>. Sulfide inclusions in weld metal are very fine as compared with the larger aggregations in ordinary steel. This is held to be due to the rapid chilling of the weld metal. Sulfur prints from slow-cooled oxyacetylene welds showed no difference in contrast between deposit and plate<sup>(H241)</sup>. Microscopic examination showed that the sulfide inclusions in the oxyacetylene weld deposits were much larger and less numerous than those in the metal-arc welds.

In unrestrained fillet-weld cracking tests, sulfur was found to increase hot cracking and porosity in weld metal deposited with AWS Class E6010, E6012, and E6020 electrodes<sup>(H60)</sup>. Hot cracking was also severe with E6015 weld deposits containing 0.16 per cent sulfur, but was absent in E6015 welds containing 0.054 per cent sulfur. Generally, sulfur had much less effect on weld metal deposited with low-hydrogen electrodes than weld metal deposited with any other electrodes<sup>(H243)</sup>. Free-machining steels were welded satisfactorily with low-hydrogen electrodes.

Weld metal deposited with AWS Class EXX20 and EXX30 electrodes were generally more susceptible to sulfur hot cracking than weld metal



deposited with any other class of electrode<sup>(H51)</sup>. In an investigation on weld-metal cracking, it was concluded that sulfur in E6020 core wire should not exceed 0.03 per cent<sup>(H243)</sup>. Even raising the manganese content of the core wire did not seem to decrease the susceptibility of the weld metal to sulfur cracking.

Sulfur caused cracking in weld metal deposited by the atomic-hydrogen welding process, using SAE 6130 steel filler wire, when present in amounts exceeding 0.01 per cent<sup>(H9)</sup>. The cracking seemed to be related to the aluminum content of the filler wire. The aluminum tended to form intergranular films of aluminum sulfide. However, if enough aluminum was added, the films appeared to coalesce into aluminum sulfide inclusions which apparently do no harm. In another investigation, the idea was advanced that the aluminum content controlled the shape of iron-manganese-sulfide inclusions in the weld metal<sup>(H123)</sup>.

### Phosphorus

Helin and Svantesson<sup>(H116)</sup> found that phosphorus, when present in weld metal up to 0.07 per cent, has little effect on weld-metal cracking. Spraragen and Claussen<sup>(H255)</sup> placed this limit at 0.15 per cent, in their literature survey of published work, prior to July, 1937, on the effect of phosphorus on the welding of steel. Phosphorus in amounts up to 0.28 per cent did not cause hot cracking or porosity in ferritic weld metal deposited with AWS Class E6010, E6012, or E6020 electrodes<sup>(H60)</sup>. However, the high-phosphorus welds fractured in a brittle manner at room temperature under a sledge. On the other hand, Stieler<sup>(H266)</sup> stated that phosphorus causes cracking in weld metal deposited with both coated and uncoated ferritic electrodes. The most crack-sensitive weld contained 0.14 per cent carbon and 0.071 per cent phosphorus.

Rollason<sup>(H215)</sup> suggested that phosphorus has an adverse effect on hot cracking because it affects the segregation of sulfur. He obtained the following results in Murex hot-cracking tests on welds made with ferritic electrodes.

Specimen	Composition, per cent					Length of Crack, inches
	C	Mn	S	P	Si	
S	0.13	0.55	0.02	0.039	0.15	0
F	0.125	0.51	0.022	0.072	0.15	1.7

Oxygen. Hot cracking in ferritic weld metal has sometimes been attributed to oxygen. Its effect appears to be due to its combination with other elements to form low-melting compounds such as FeO. Goodger<sup>(H104)</sup> concluded that oxygen caused red shortness in weld metal deposited with bare-wire electrodes. He observed that fracture failures in heated specimens followed brittle oxide films in the grain boundaries though this occurrence can be otherwise interpreted. The hot shortness was removed by heating in a reducing atmosphere or by adding deoxidizing agents. However, it should be pointed out that cracking during welding could be due to factors other than hot shortness, i. e., hot tearing, blue brittleness. Stieler<sup>(H266)</sup> stated that welds deposited with bare-wire electrodes were less crack sensitive than welds deposited with coated electrodes (no mention of coating type), which would seem to indicate that oxygen might not be a factor in cracking. Reeve<sup>(H201)</sup> also found that the least oxidized weld metal, as measured by FeO content, was most sensitive to hot cracking. Table 7 shows the analysis obtained by Reeve on various weld-metal deposits. These are listed in the order of decreasing FeO content.

Rollason<sup>(H215)</sup> believed that under certain conditions FeO in the weld metal actually reduced the effect of sulfur on hot cracking. Up to about 0.06 per cent sulfur, it was considered that FeS combined with FeO and thus had less effect on grain-boundary weakening. This was based on his observation that deoxidation with a small amount of manganese increased cracking. However, he pointed out that higher sulfur would cause formation of FeS-FeO eutectic films which would increase cracking susceptibility. Reeve<sup>(H201)</sup> also felt that FeO might reduce cracking by combining with FeS. He observed, microscopically, that in crack-free welds a greater quantity of sulfur combined with the larger oxide globules than was the case in cracked welds. Hoyt<sup>(H132)</sup> attributed shrinkage cracks in high-oxygen weld metal to finely dispersed FeO. He concluded that weld metal deposited with bare-wire electrodes was hot short as a result of bend tests at 1920°F.

Hydrogen. The possible influence of hydrogen on hot cracking in weld metal is not very clear. Stieler<sup>(H266)</sup> found that the least crack sensitive of a series of ferritic welds contained three times as much hydrogen as the others, and therefore concluded that hydrogen had no effect. Reeve<sup>(H201)</sup> suggested that the formation of hydrogen sulfide gas in increasing quantities, as the temperature of ferritic weld metal dropped from a red heat, might play some part in producing cracks by internal gas pressure. However, he also pointed out that the most oxidized electrodes were also lowest in diffusible hydrogen as shown in Table 7. Helin and Svantesson<sup>(H116)</sup> stated that hydrogen pressures might become equal to and even exceed the cohesive forces of still-pasty weld metal. They suggested that the sharp decrease in solubility of hydrogen on transformation from austenite to ferrite might contribute to cracking in this temperature range, but this transformation must occur at a relatively low temperature.

TABLE 7. OXYGEN, NITROGEN, AND HYDROGEN CONTENT OF WELDS(H201)

Electrode Type	All-Weld-Metal Analysis										Oxygen		Nitrogen,		Hydrogen	
	C, per cent	P, per cent	Mn, per cent	S, per cent	Si, per cent	FeO, per cent	MnO, per cent	SiO <sub>2</sub> , per cent	Al <sub>2</sub> O <sub>3</sub> , per cent	Total O <sub>2</sub> , per cent	Total per cent	Diffusible cc/100 g	Vac. Fusion cc/100 g			
K <sub>1</sub> Dead soft	0.04	0.025	0.02	0.024	Trace	0.859	0.013	0.0035	0.0095	0.200	0.059	1.53	10.2			
K <sub>2</sub> Dead soft	0.04	-	0.02	-	Trace	-	-	-	-	-	-	Trace	-			
L Dead soft	0.03	-	0.02	-	Trace	0.916	0.015	0.0055	0.0015	0.211	0.077	Trace	9.2			
M Dead soft	0.03	0.025	Trace	-	Trace	-	-	-	-	-	-	Trace	-			
A Soft	0.04	0.034	0.14	0.036	Trace	0.723	0.0375	0.0095	0.0045	0.1765	0.087	2.0	7.4			
B Soft Class A	0.07	0.025	0.33	0.034	0.028	0.181	0.189	0.055	0.075	0.148	0.020	1.28	13.6			
H <sub>1</sub> Class A, General purpose	0.05	0.045	0.32	0.035	0.056	0.086	0.106	0.0165	0.013	0.058	0.053	-	6.1			
C Class A, General purpose	0.06	0.028	0.33	0.035	0.085	0.068	0.127	0.048	0.041	0.089	0.031	2.02	10.0			
D Class A, Butt welds	0.09	0.041	0.39	0.029	0.038	0.046	0.151	0.085	0.0125	0.095	0.044	3.85	17.6			
G Class A, General purpose	0.05	0.035	0.40	0.023	0.132	0.036	0.122	0.032	0.011	0.0575	0.0425	6.5	16.0			
F Class A, Butt welds	0.08	0.030	0.29	0.040	0.118	0.024	0.124	0.042	0.020	0.0655	0.034	4.95	13.2			

There is considerable evidence that hydrogen can cause low ductility, fissuring, and cold cracking in ferritic weld metal. It is possible that hydrogen plays the same role in weld-metal cracking as it does in causing underbead cracking in the base metal. An example of underbead cracking was shown in Figure 16 earlier in the report. Many investigators have reported that underbead cracking originates from hydrogen in the arc atmosphere. During welding, the weld metal dissolves a large amount of hydrogen, which diffuses into the base metal adjacent to the weld. Some of the hydrogen is believed to remain in retained austenite in the overheated portion of the base-metal heat-affected zone.

When this austenite, which is unstable at ordinary temperatures, finally transforms, it releases the hydrogen. The hydrogen in the monatomic form then diffuses to microscopic or submicroscopic rifts in the base metal where it then combines as molecular hydrogen. The aerostatic pressure in these microscopic rifts can sometimes develop to a point where rupture of the metal initiates the under-bead cracks. These cracks are then propagated by thermal stresses in the weld zone. Since this survey was concerned with weld-metal cracking, an extended discussion of underbead cracking will not be included here. However, a bibliography of 141 references on under-bead cracking collected during the survey is included in Appendix J.

As previously discussed, Miller and Kovac<sup>(H182)</sup> reported the occurrence of fissures in low-carbon-steel weld metal. They concluded, however, that hydrogen did not cause the fissures because they were intergranular and occurred only in recrystallized zones which had been reheated during deposition of subsequent weld beads. They assumed that intergranular cracks could occur only above the equicohesive temperature of steel. This assumption is not necessarily valid, since it has been shown, for example, that intergranular cracks can occur at room temperature in cast steel<sup>(A23)</sup>.

The more recent work of Flanigan, et al<sup>(H93, H94, H95)</sup>, Bland<sup>(H37)</sup>, and Rollason and Roberts<sup>(H214)</sup> shows that intergranular fissuring in arc welds in mild steel is probably related to hydrogen. Flanigan<sup>(H93)</sup> produced low ductility in E6010 weld metal by water quenching after completion of welding. He observed that the low ductility was caused by microcracks. A typical microcrack was shown in Figure 25. He further demonstrated the influence of hydrogen by showing that the introduction of hydrogen by pickling of weld-metal samples produced an embrittling effect; and also showed that post heating improved the ductility of the E6010 weld metal. Post-heating temperatures ranged from 250° to 1150°F with exposure times of 1/2 hour to more than 100 hours.

Later work by Flanigan, et al<sup>(H94, H95)</sup>, indicated that cooling rates might be important at temperatures as low as 200°F; this further supports

the hydrogen idea. These investigators also found that low-hydrogen electrode weld deposits were immune to embrittlement by rapid cooling at low temperatures.

Bland<sup>(H37)</sup> showed that quenching E6010 welds in boiling water immediately after welding caused low ductility in the weld metal. He found that quench times of 3 minutes or longer were required for maximum ductility. "Quench time" was the lapse of time between completion of welding and the final quenching in the boiling water. The results indicated that the amount of hydrogen in the weld deposit prior to quenching was the critical factor.

Steinberger, et al<sup>(H264)</sup>, concluded that cold cracks in circular-groove atomic-hydrogen welds made with various alloy-steel filler wire were related to hydrogen. The filler metals studied were SAE 4320, 4330, 6130, and 4130; N. E. 8630 low alloy steels and 310, 347, and 410 stainless steels. All of the weld-metal deposits made with these filler metals developed cracks when air-blast cooled, except the austenitic welds made with Type 310 and 347 filler wires. In discussing the results of this investigation, Sims<sup>(H238)</sup> agreed with the authors that hydrogen was a cause of the cracking in the high-strength weld metal.

Aluminum. Aluminum is used as a deoxidizer and grain refiner in steel, and has been added to filler wire to reduce porosity in ferritic welds made by the atomic-hydrogen welding process. Several investigators<sup>(H9, H224)</sup> have suspected aluminum to be a cause of intergranular cracking in atomic-hydrogen welds made in SAE 4320 steel with SAE 6130 steel filler wire. There was some evidence that small amounts of aluminum might combine with sulfur to form intergranular films of sulfide which could lead to cracking<sup>(H9)</sup>. However, if excess aluminum was added, the aluminum sulfide seemed to coalesce into a less harmful form.

Nickel. Nickel is reported to increase the susceptibility of ferritic weld metal to sulfur hot cracking<sup>(H215)</sup>. The effect of the nickel may be related to the low-melting point of NiS (1465°F). The nickel plus sulfur could possibly form low-melting films in the grain boundaries. Rollason<sup>(H215)</sup> obtained the following data on Murex weld-cracking tests to show the effect of nickel on welds deposited with E 317 (Rutile-type coating) electrodes.

<u>Wire, Nickel Content,</u> <u>per cent</u>	<u>Cracking Length</u> <u>with Nickel, inches</u>
0.08	1/2
0.32	1-3/4
0.60	3

Vanadium. Vanadium is added to steel to refine the grain structure and to stabilize carbides at elevated temperatures. It tends to strengthen weld metal. As previously mentioned when discussing the effects of carbon, the vanadium, when added to highly hardenable weld metal, is said to reduce cold cracking by forming stable carbides which are but slowly soluble in austenite(H238, H264).

Intergranular cracking in SAE 6130 steel weld metal (refer to Figure 28) seemed to have a possible relation to the vanadium and aluminum content of the weld metal(H224). Although vanadium is a grain refiner, it may also be conducive to excessive grain coarsening when the weld metal is overheated (as in multipass welds). The formation of carbides and nitrides of vanadium at the grain boundaries was suspected, but no visual evidence was obtained(H224).

Nitrogen. Nitrogen can be retained in ferritic weld metal partly as iron nitride, ( $Fe_4N$ ) or as other nitrides, and partly as dissolved nitrogen (I23). Goodger(H104) reported that iron will take up as much as 7 per cent by weight when heated in ammonia gas. This percentage seems excessively high as compared with other evidence. Reeve(H205) reported nitrogen contents of 0.16 per cent in bare-wire welds, and 0.05 per cent in welds made with coated electrodes. Hoyt(H132) found 0.010 per cent nitrogen in a deposit made with cellulose-covered electrodes. Iron nitride appears in iron as needles separated out along the cleavage planes. The iron nitride was believed to cause embrittlement(A4).

It has not been definitely established that nitrogen contributes to weld-metal cracking. Helin and Svantesson(H116) and Reeve(H201) concluded that nitrogen had no noticeable effect on hot cracking in mild-steel weld metal. Some early investigators suggested that hot shortness in ferritic weld metal deposited with bare-wire electrodes was due to nitrogen. However, Goodger(H104) concluded that oxygen was the cause rather than nitrogen.

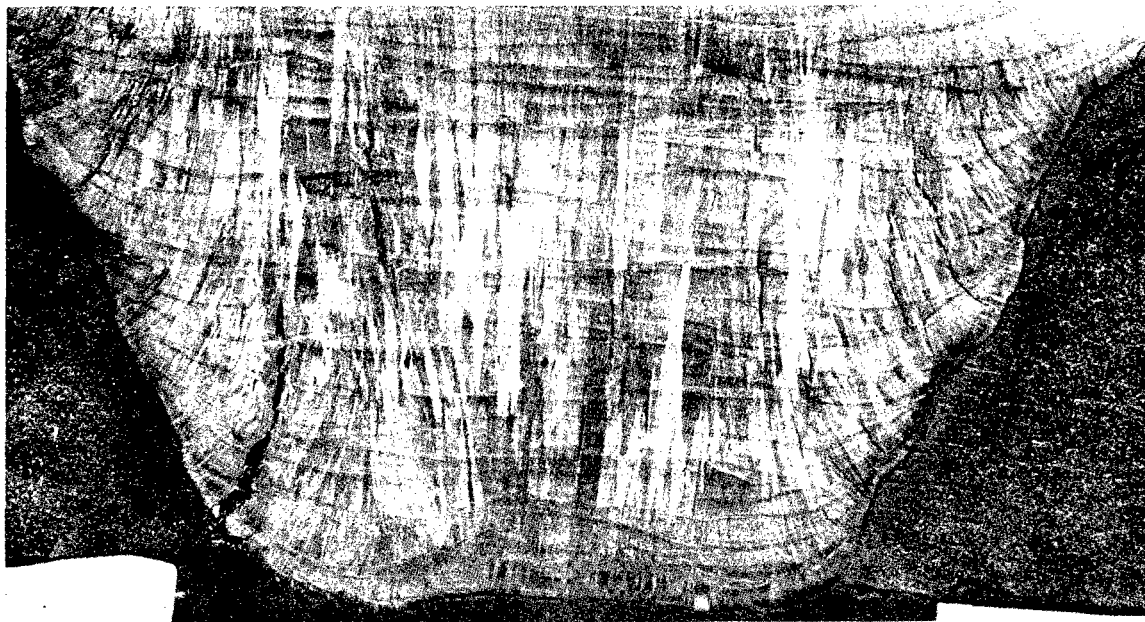
Gaymans(H102) suggested that blue brittleness, which is related to carbon and nitrogen in the crystal lattice, might be a cause of weld-metal cracking that occurs at 1100°F or lower.

A number of early investigators(H125, H132, H193, H194, H195, H205) have reported that nitrogen causes brittleness in welds, but none of these have suggested that this brittleness might be related to weld-metal cracking.

### Austenitic Weld Metals

The greater susceptibility of austenitic weld metal to cracking may be related to its strong tendency to form large columnar grains or dendrites.

In multipass welds, the dendrites are generally not confined to a single bead, but may extend through subsequent bead deposits. The boundaries of these large dendrites are a possible zone of weakness and may often contain microfissures and cracks, as shown in Figure 32. The susceptibility of austenitic weld metal to this type of interdendritic cracking varies with chemical composition.



4X

Ferric Chloride Etch

FIGURE 32. CRACKS IN THE BOUNDARIES OF LARGE COLUMNAR DENDRITES OF TYPE 310 WELD METAL(I11)

Delta Ferrite. Weld-metal deposits made with chromium-nickel stainless-steel electrodes are generally termed "austenitic" because their microstructure is predominately austenite at room temperature. Tests by various investigators have indicated that austenitic weld metal containing a small amount of delta ferrite\* in the structure is less crack sensitive than a purely austenitic weld metal. The part played by delta ferrite in reducing hot cracking was probably first reported by Riedrich<sup>(I33)</sup>, in Germany, who stated that a certain proportion of ferrite in austenitic electrodes was beneficial in reducing "shrinkage" cracking in austenitic weld-metal deposits.

The delta-ferrite content of chromium-nickel weld deposits can be controlled by composition. It is known that chromium and molybdenum tend to induce the formation of delta ferrite in the microstructure of chromium-nickel steels; and that elements such as nickel, manganese, and carbon suppress its formation(I12, I15, I38). To estimate the microstructure from composition, the following formula was first suggested by

\*Delta ferrite is the first phase to form during solidification of low-carbon steel. It has a body-centered-cubic lattice structure.

Newell and Fleishman<sup>(I28)</sup>, and was later modified by Field, Bloom, and Linnert<sup>(I8)</sup> to apply to austenitic chromium-nickel weld-metal deposits:

$$\text{Ni} < \frac{(\text{Cr} + 2 \text{ Mo} - 16)^2}{12} - \frac{\text{Mn}}{2} + 30 (0.10 - \text{C}) + 11. \quad (1)$$

To produce ferrite in the microstructure, the percentage of nickel in an austenitic stainless steel electrode should be less than the value obtained from the formula. If the nickel content is higher, the structure will probably be entirely austenitic.

The weld-metal-structure formula was further revised as a result of many additional examinations of weld-metal deposits<sup>(I40)</sup>.

$$\text{Ni} = 1.1 (\text{Cr} + \text{Mo} + 1.5 \text{ Si} + 0.5 \text{ Cb}) - \frac{\text{Mn}}{2} - 30\text{C} - 8.2. \quad (2)$$

Schaeffler<sup>(I35)</sup> constructed the diagram shown in Figure 33 from this formula. This diagram has been helpful to users of stainless steel electrodes in balancing weld-metal compositions.

The photomicrographs in Figure 34 show the influence of base-metal composition on delta ferrite in the microstructure of Type 316 (18 Cr - 12 Ni - 3 Mo) weld-metal deposits. The nickel content of the 19-9 DL alloy (19 Cr - 9 Ni - 1 W - 1 Mo) was lower than for Timken Alloy (16 Cr - 25 Ni - 6 Mo); this apparently favored the formation of delta ferrite in the microstructure of the weld on the 19-9 DL alloy.

Emerson and Morrow<sup>(I7)</sup> found microcracks and fissures in submerged-arc weld deposits made with 19-9 Cb electrodes. These welds contained only 2 per cent ferrite; and furthermore, the ferrite was not uniformly dispersed. The cracks were found in the areas of least ferrite content.

A high percentage of delta ferrite in weld metal is not always desirable because, if such a weld is heated to high temperatures in service, a transformation to the brittle sigma phase may take place. As the brittle nonmagnetic sigma phase forms, the hardness of the weld metal increases and its ductility is drastically reduced. The transformation of delta ferrite to sigma phase in heat treated Type 349 (19 Cr - 9 Ni - W - Mo) weld-metal deposits is illustrated in Figure 35. Emerson and Morrow<sup>(I7)</sup> found that the ferrite-to-sigma phase change in austenitic alloys high in ferrite was 8.5 per cent complete in 300 hr at 1100°F. However, they also found that ferrite, not sigma, was the stable phase at temperatures of 1800°F and higher.

Where austenitic weld metal is to be used under extremely corrosive conditions, choice of the correct amount of delta ferrite is a problem. Schaeffler and Thomas<sup>(I37)</sup> stated that to prevent fissures or hot tears,



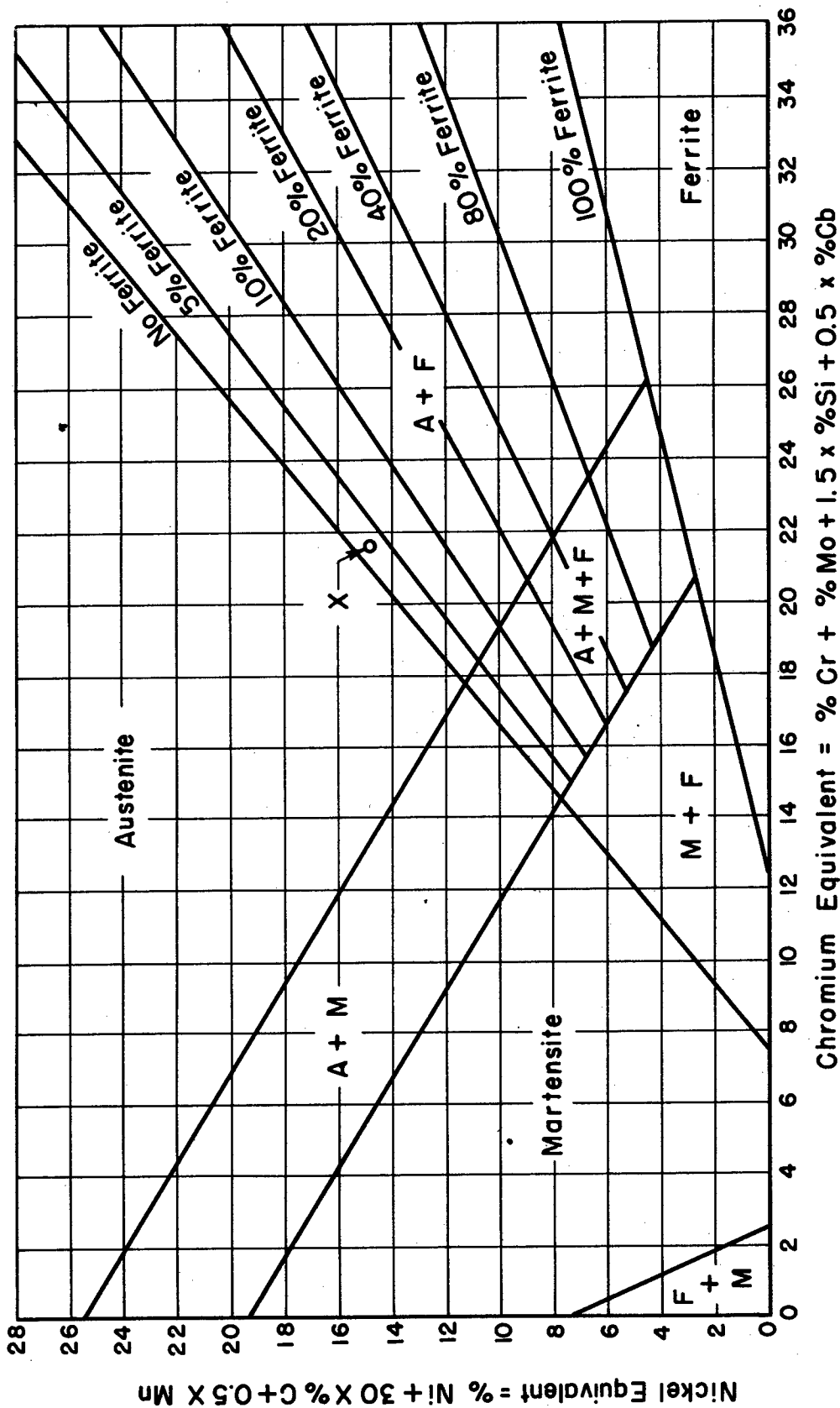
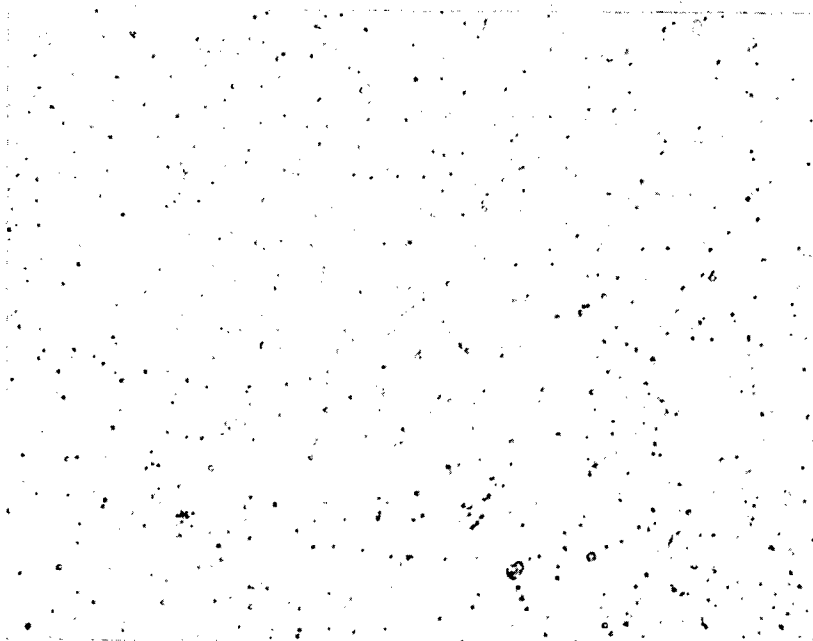


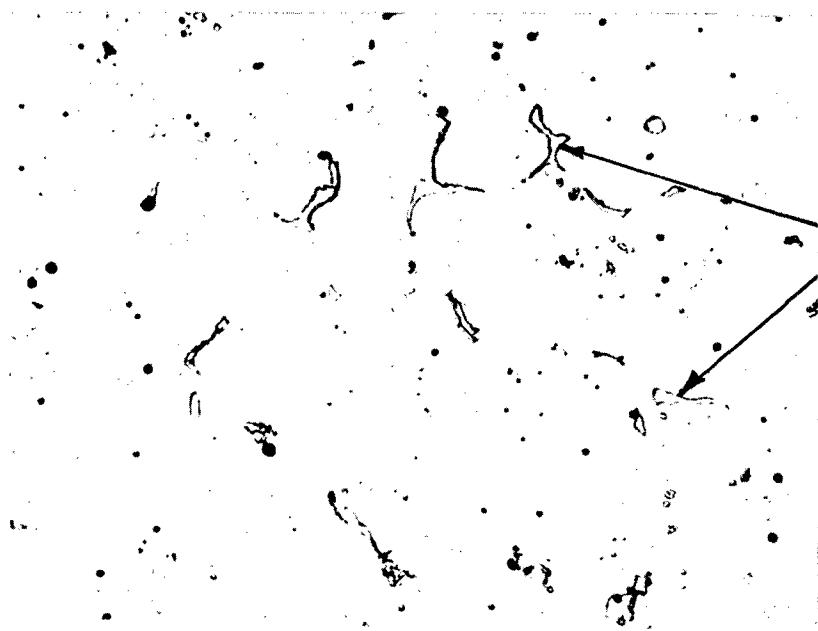
FIGURE 33. DIAGRAM FOR ESTIMATING THE MICROSTRUCTURE IN STAINLESS STEEL WELD METAL (I35)



1000X

Aqua Regia Etch

A. Fully austenitic



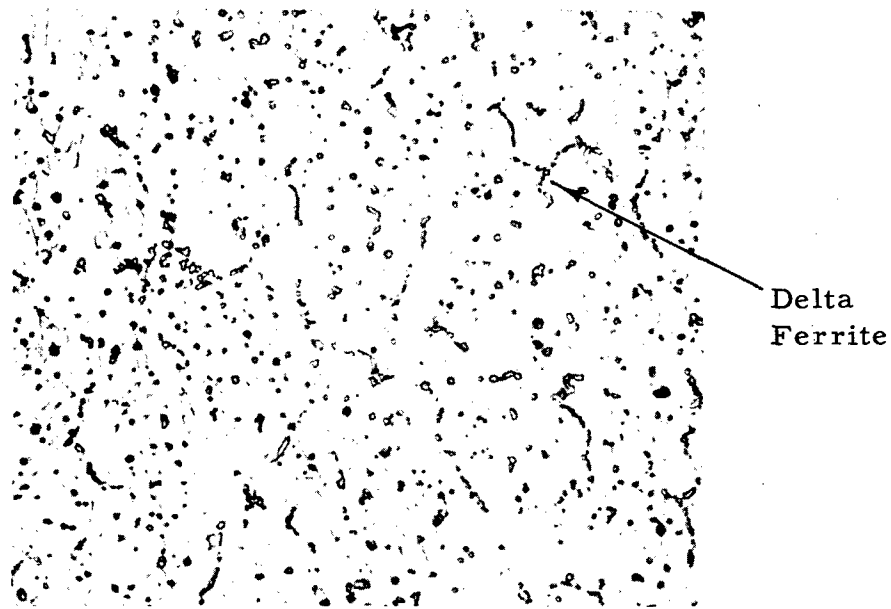
1000X

Aqua Regia Etch

Delta Ferrite

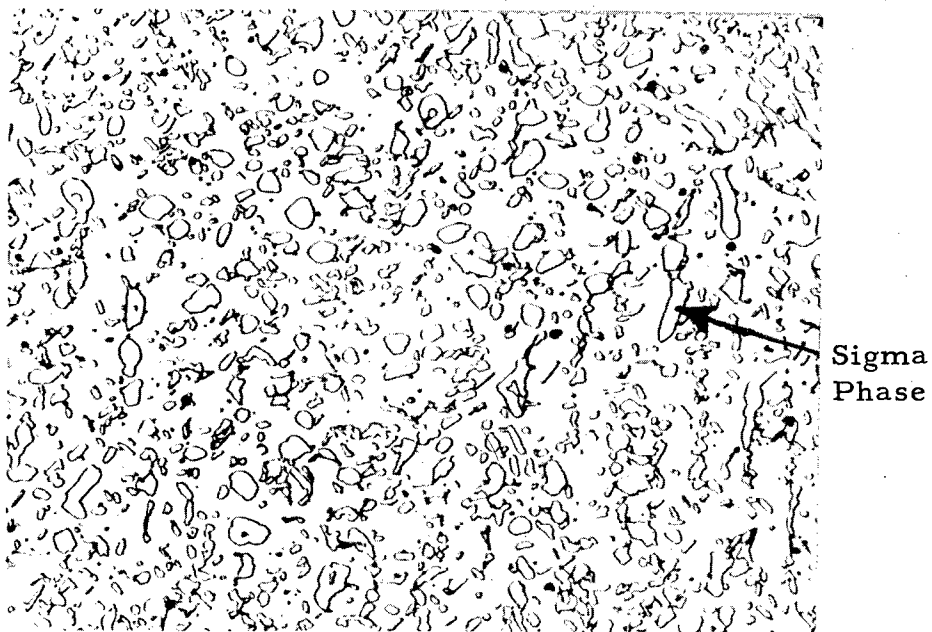
B. About 3 per cent delta ferrite

FIGURE 34. MICROSTRUCTURES OF TWO WELD-METAL DEPOSITS MADE WITH TYPE 316 ELECTRODES ON: A. TIMKEN 16-25-6 ALLOY; AND B. 19-9 DL ALLOY, SHOWING INFLUENCE OF BASE-METAL COMPOSITION ON INCIDENCE OF DELTA FERRITE



1000X Aqua Regia Etch

A. Water quenched, 2000°F



1000X Aqua Regia Etch

B. Water quenched, 2000°F, aged 120 hrs at 1560°F

FIGURE 35. MICROSTRUCTURES SHOWING EFFECT OF HEAT TREATMENT ON TRANSFORMATION OF DELTA FERRITE TO SIGMA IN WELD METAL DEPOSITED WITH TYPE 349 ELECTRODES

delta ferrite is desirable and should exceed 4 per cent in austenitic weld metal for maximum resistance to nitric acid corrosion. Excessive delta ferrite tends to transform to sigma, causing a significant loss in ductility.

In a recent investigation of hot ductility in unwelded stainless steel, it was shown that ferrite decreased hot ductility of austenite as determined by hot-twist tests at elevated temperatures<sup>(11)</sup>. It was found that a small amount of austenite in a ferritic alloy also reduced hot ductility. The authors stated that two-phase structures were not so ductile as the pure phases. This seems to be converse to the idea that delta ferrite in austenitic weld metal is beneficial in reducing hot cracking.

Thomas<sup>(145)</sup> grouped the following weld metals in the order of increasing crack susceptibility and made some very interesting comments about each type of weld metal.

<u>Type 301 (16 Cr - 7 Ni)</u>	Least susceptible to cracking; high in ferrite content.
<u>Type 308 (19 Cr - 9 Ni)</u>	Medium ferrite content.
<u>Type 347 (19 Cr - 9 Ni-Cb)</u>	The increase in ferrite by Cb is usually balanced by Ni additions. Cb increases the crack susceptibility of austenitic weld metal except when it increases the delta-ferrite content.
<u>Type 316 (18 Cr - 13 Ni-Mo)</u>	Sufficient nickel is added to counterbalance the ferrite-forming tendency of Mo. This is the least crack sensitive of the fully austenitic alloys.
<u>Type 309 (25 Cr - 12 Ni)</u>	Fully austenitic.
<u>Type 309 + Cb (25 Cr - 12 Ni-Cb)</u>	Cb tends to make the structure partially ferritic, which counterbalances cracking tendency otherwise caused by it.
<u>Type 310 (25 Cr - 20 Ni)</u>	Fully austenitic.
<u>Type 310 + Cb (25 Cr - 20 Ni-Cb)</u>	Cb cannot form ferrite in this alloy, and, therefore, makes it more crack sensitive than ordinary Type 310.
<u>Type 330 (15 Cr - 35 Ni)</u>	Fully austenitic alloys having more Ni than Cr are more sensitive to cracking than those with more Cr.

It is interesting to observe that cracking susceptibility increases with the higher alloy weld metals, and that the ratio of chromium to nickel is important. It is also interesting that columbium increases the cracking tendency, unless it produces delta ferrite in the right amount.

The mechanism by which delta ferrite reduces the weld-metal cracking tendency is not known. It may be that the ferrite constituent, being present in a finely dispersed pattern, is not favorable to the formation of cracks<sup>(I45)</sup>. It has also been suggested that the delta ferrite increases the grain-boundary area and thus decreases the width of the grain boundary<sup>(I23)</sup>. It is generally recommended that the delta-ferrite content of the weld metal exceed 5 per cent to prevent fissuring and hot cracking in chromium-nickel steel weld deposits<sup>(I8, I43, I45)</sup>.

Silicon. A certain amount of silicon is necessary in austenitic weld metal, because of the poor physical properties and unsatisfactory arc and slag characteristics which would result with incomplete deoxidation. However, silicon in excessive amounts tends to lower the strength and ductility and also increases the tendency to fissuring. The detrimental effect of silicon can be minimized somewhat by adding carbon.

Campbell and Thomas<sup>(I3)</sup> concluded that excessive silicon caused fissuring in weld metal deposited with Type 310 (25 Cr - 20 Ni) electrodes. They also found that silicon and phosphorus were additive, and recommended that silicon be limited to 0.60 per cent maximum when phosphorus is 0.016 per cent, but should be lowered to 0.40 per cent maximum when the phosphorus content is 0.024 per cent.

Weld-metal hot cracking in manual-arc welds in nickel-chromium-iron alloys increased with the amount of silicon in the weld metal<sup>(I17)</sup>. An addition of columbium seemed to counteract the effect of silicon. A critical value for the Cb/Si ratio to prevent hot cracking was determined for each alloy as follows:

<u>Alloy</u>	<u>Cb/Si Ratio</u>
80 Ni - 20 Cr	3.8
80 Ni - 14 Cr - 6 Fe	4.5
60 Ni - 15 Cr - 25 Fe	7.0 (Approx.)
35 Ni - 15 Cr - 50 Fe	8.0 (Approx.)

The authors<sup>(I17)</sup> used the Cb/Si principle to develop an electrode that produced sound welds on the 80-14-6 alloy, with excellent physical properties in plate up to 1 in. in thickness.

Carpenter and Jessen<sup>(15)</sup> hypothesized that the harmful effects of high silicon in Type 310 (25 Cr - 20 Ni) weld-metal deposits were not due to metallic silicon in the austenite, but rather to the formation of silicate films surrounding the austenite grains. They based this on metallographic evidence of intergranular slag films in the weld metal as shown in Figure 26. They found that silicon in excess of 0.50 per cent caused defects in low-carbon (below 0.12 per cent C) weld metal. At higher carbon levels (0.18 - 0.20 per cent C), silicon up to approximately 0.70 per cent was allowable. In a discussion of this paper, Campbell and Thomas<sup>(14)</sup> stated that differences in phosphorus contents of the high-silicon weld metals probably had more effect on the results than the differences in silicon content.

Silicon and carbon were suspected to cause reduced strength and ductility in weld metal deposited with 19 Cr - 9 Ni electrodes<sup>(16)</sup>. Non-metallic intergranular films similar to those observed in Type 310 (25 Cr - 20 Ni) weld metal were believed to cause the cracking<sup>(16)</sup>. The results of tensile tests showing the relationship between the weld-metal properties and the silicon and carbon contents are given in Table 8. In discussing this table, Linnert<sup>(123)</sup> felt that the chromium-nickel ratio was a more important factor than the carbon and silicon. He stated that the high Cr/Ni ratio of the last four weld-metal compositions (E-646, E-647, E-648, E-649) indicated that they contained some delta ferrite in their microstructure, which accounted for their freedom from fissuring.

Some recent difficulties were encountered in welding Type 314 (25 Cr - 20 Ni-Si) steel using Type 314 electrodes<sup>(123)</sup>. Because of the high silicon content, the welds were extremely hot short and were highly susceptible to cracking<sup>(123)</sup>. The cracking was finally eliminated by using either Type 312 (29 Cr - 9 Ni) or Type 349 (19 Cr - 9 Ni-W-Mo) electrodes. This illustrates the suppression of the effects of silicon by the use of a weld metal containing delta ferrite.

In an investigation on the effects of composition on the strength of Type 330 (15 Cr - 35 Ni), weld-metal cracking was found with high silicon if the carbon content was below 0.40 per cent<sup>(134)</sup>. Figure 36 shows this for Type 330 steel weld metal for the range of 0.16 - 0.25 per cent carbon. As mentioned in discussing the effects of carbon, silicon without sufficient carbon causes fissuring in Type 347 (19 Cr - 9 Ni - Cb) weld-metal deposits also.

Silicon was also a suspected cause of hot cracking in welds deposited on 2 per cent silicon (ferritic) steel using Type 316 (25 Cr - 12 Ni) stainless steel electrodes. Crack-free welds were finally made with Type 310 (25 Cr - 20 Ni) electrodes<sup>(H164)</sup>. The higher carbon content (0.21 per cent) of the Type 310 weld metal compensated for the pickup of silicon in the weld metal.

TABLE 8. EFFECTS OF CARBON AND SILICON ON THE PROPERTIES OF FULLY AUSTENITIC  
19 Cr - 9 Ni WELD METALS IN THE AS-WELDED CONDITION<sup>(1)</sup>

Specimen No.	Yield Point, psi	Tensile Strength, psi	Elong., per cent	Red. of Area, per cent	Composition, per cent							Cr/Ni <sup>(1)</sup> Ratio	Fissures
					C	Mn	Si	Cr	Ni	S	P		
E-121	54,000	70,750	13.0	27.8	0.09	2.19	0.35	17.74	13.99	0.015	0.015	1.27	Many
E-122	57,500	79,000	28.5	44.5	0.09	2.27	0.38	18.30	11.64	0.017	0.011	1.57	Few
E-654	38,250	58,640	23.0	44.7	0.07	1.92	0.75	18.03	12.64	0.010	0.012	1.43	Many
E-645	35,500	73,000	36.5	42.8	0.10	2.05	0.85	17.97	11.96	0.012	0.014	1.50	Many
E-646	45,500	91,000	43.0	57.0	0.17	2.04	0.41	18.43	8.92	0.012	0.012	2.07	None
E-647	51,500	82,250	53.0	62.2	0.18	2.02	0.48	18.74	9.12	0.016	0.012	2.05	None
E-648	49,000	80,000	47.0	64.4	0.17	2.14	0.62	18.63	9.48	0.015	0.012	1.97	Very few
E-649	48,000	79,000	41.5	39.8	0.18	2.06	0.70	18.21	9.40	0.012	0.016	1.94	Few

(1) Not included in original table.

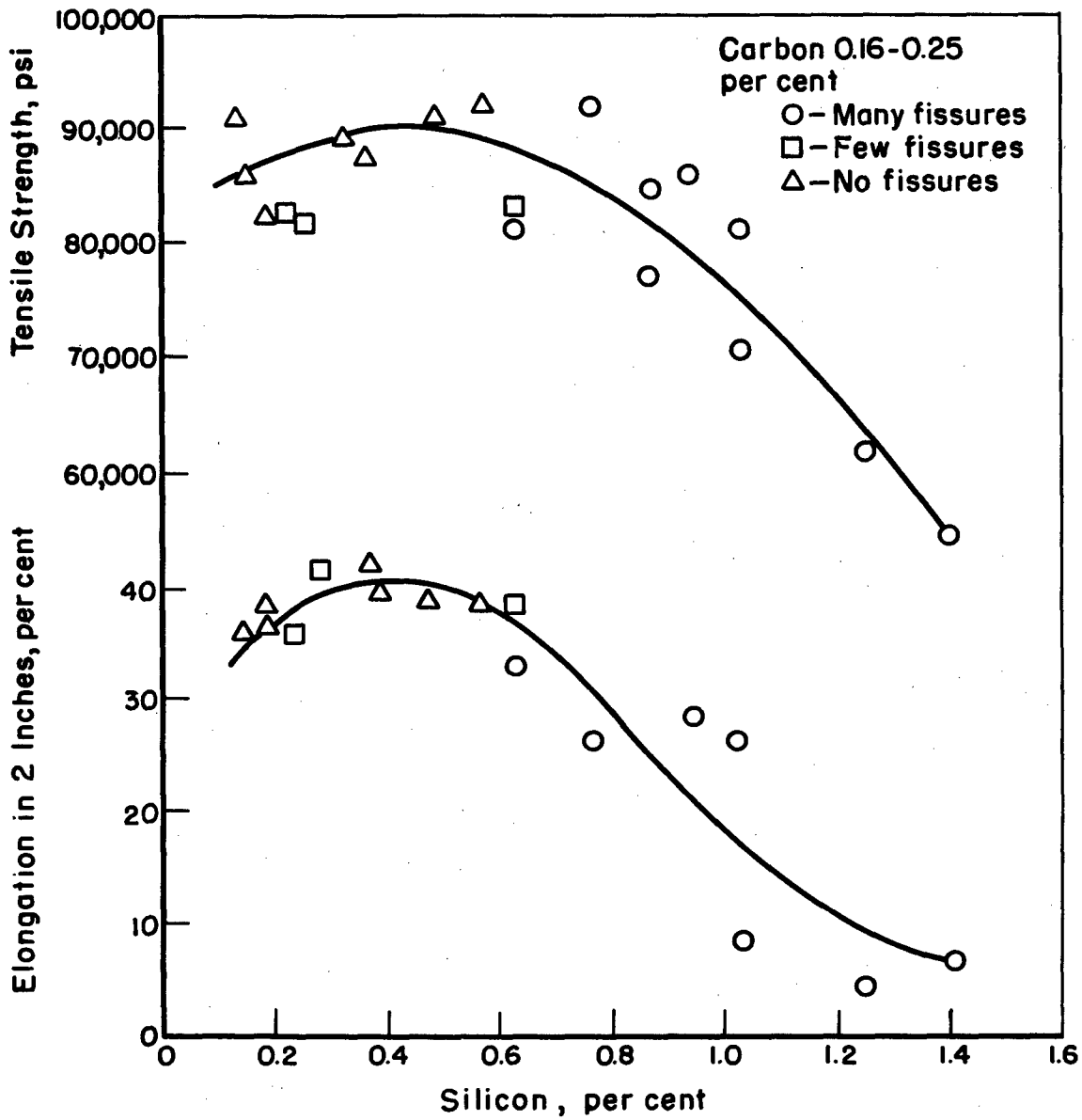


FIGURE 36. EFFECT OF SILICON ON ELONGATION AND TENSILE STRENGTH OF WELD METAL DEPOSITED WITH TYPE 330 STEEL ELECTRODES (I 34)



Columbium. Columbium is added to austenitic steel electrodes to stabilize the weld metal. Although columbium promotes the formation of ferrite, it nevertheless tends to increase the susceptibility of fully austenitic weld metal to hot cracking.

Campbell and Thomas<sup>(I3)</sup> concluded that fissuring was eliminated in Type 310 (25 Cr - 20 Ni) weld metal containing excess silicon because the columbium addition introduced the ferrite phase. However, Type 310 weld metal with a fully austenitic microstructure is extremely sensitive to cracking when columbium is added<sup>(I45)</sup>.

As previously mentioned, columbium was found to counteract the deleterious effects of silicon in causing cracking in welds made with several nickel-chromium-iron alloys<sup>(I17)</sup>. Fissuring in Type 330 (15 Cr - 35 Ni) weld metal containing about 0.10 per cent carbon was eliminated by adding columbium<sup>(I34)</sup>, whereas an increase in carbon to 0.20 per cent would have had the same effect.

Crater cracks in weld metal deposited with Type 302 (19 Cr - 9 Ni) electrodes were attributed to additions of columbium<sup>(I6)</sup>. They stated that, "The cracks appeared to be related to an intergranular constituent and their size seemed to be directly proportional to the columbium content". Similar cracks were also observed in 25 Cr - 20 Ni-Cb steel, 16 Cr - 13 Ni - 3 Mo - Cb steel, carbon steel to which columbium was added, and Armco iron weld-metal deposits<sup>(I6)</sup>. The author felt that if no carbon were present, the crater cracking would not occur even with 1.00 per cent columbium<sup>(H51, I6)</sup>. Other investigators<sup>(I23, H278)</sup> disagreed with this assumption, since their work has indicated that an extremely low carbon content increases the cracking tendency.

In general, the effect of columbium in reducing hot cracking in some austenitic weld metals is not easily explained, considering that columbium is known to cause hot shortness<sup>(H164)</sup>. Cracks have been observed in S-590 (Ni-Cr-Co-Fe) alloy, after bending at 1000°F, which seemed to follow the columbium carbides in the microstructure.

Molybdenum. Molybdenum is added to austenitic stainless steel for added resistance to certain types of corrosion. The cracking tendency of austenitic weld metal is reduced somewhat by adding molybdenum, since it is a ferrite former and tends to strengthen the austenite. Thomas<sup>(I45)</sup> stated that molybdenum-modified 19 Cr - 9 Ni electrodes were used for welding armor plate because of their great resistance to cracking during fabrication. The weld-metal structure contained a high percentage of ferrite. He also pointed out that even the fully austenitic Type 316 (18 Cr - 13 Ni-Mo) weld metal was highly resistant to cracking. Molybdenum also reduced the cracking susceptibility of austenitic 20 Cr - 10 Ni steel weld

metal<sup>(18)</sup>. Molybdenum additions to austenitic electrodes used in welding armor were recommended to compensate for a low chromium-nickel ratio.

Generally, molybdenum-bearing austenitic weld metals are less susceptible to hot-short cracking than balanced compositions of weld metal containing no molybdenum<sup>(137)</sup>. However, a combination of molybdenum and columbium seems to increase the tendency to hot tearing and cracking.

Phosphorus. Phosphorus is undesirable in austenitic weld metal because it tends to promote hot cracking and fissuring. Its effects depend somewhat on the balance of other elements.

As previously discussed, silicon and phosphorus both appeared to affect the tensile properties of weld metal deposited with Type 310 (25 Cr - 20 Ni) electrodes. Campbell and Thomas<sup>(13)</sup> derived an empirical formula for the phosphorus equivalent:

$$\text{Phosphorus equivalent} = P + 0.04 (\text{Si} - 0.25).$$

A graph of per cent elongation versus phosphorus equivalent of three separate heats of steel was a fairly close approximation to a straight line. The graph of tensile strength plotted against phosphorus equivalents showed a characteristic falling off in strength beyond 0.03 per cent. The graphs of tensile strength and elongation against phosphorus equivalents are shown in Figure 37. These results agree very well with those obtained by Linnert and Bloom<sup>(124)</sup> on Type 310 weld metal. It appears that the effect of phosphorus on microfissuring was less with "lime type" coated electrodes than with "titania type" electrodes (I3, I24, I45). It was found that practically all of the phosphorus in the coating goes into the weld metal<sup>(123)</sup>.

Type 307 (19 Cr - 9 Ni-Mn) and Type 308 (20 Cr - 10 Ni) weld-metal deposits showed an increase in crack sensitivity with phosphorus content<sup>(120)</sup>. There was some indication that the cracking was related to the combined effect of carbon and phosphorus, since the carbon content increased along with the phosphorus content.

Phosphorus in excess of 0.025 per cent caused weld-bead cracks and fissures to appear in weld metal deposited with Type 330 (15 Cr - 35 Ni) electrodes<sup>(134)</sup>. The effect of phosphorus was similar to the effect due to sulfur.

Sulfur. Sulfur appears to have little influence on hot cracking in austenitic weld metal unless the amounts are well above specification limits<sup>(145)</sup>. It is known that sulfur contributes to hot shortness in stainless steels, and therefore would be expected to cause weld-metal cracking if present in excessive amounts.

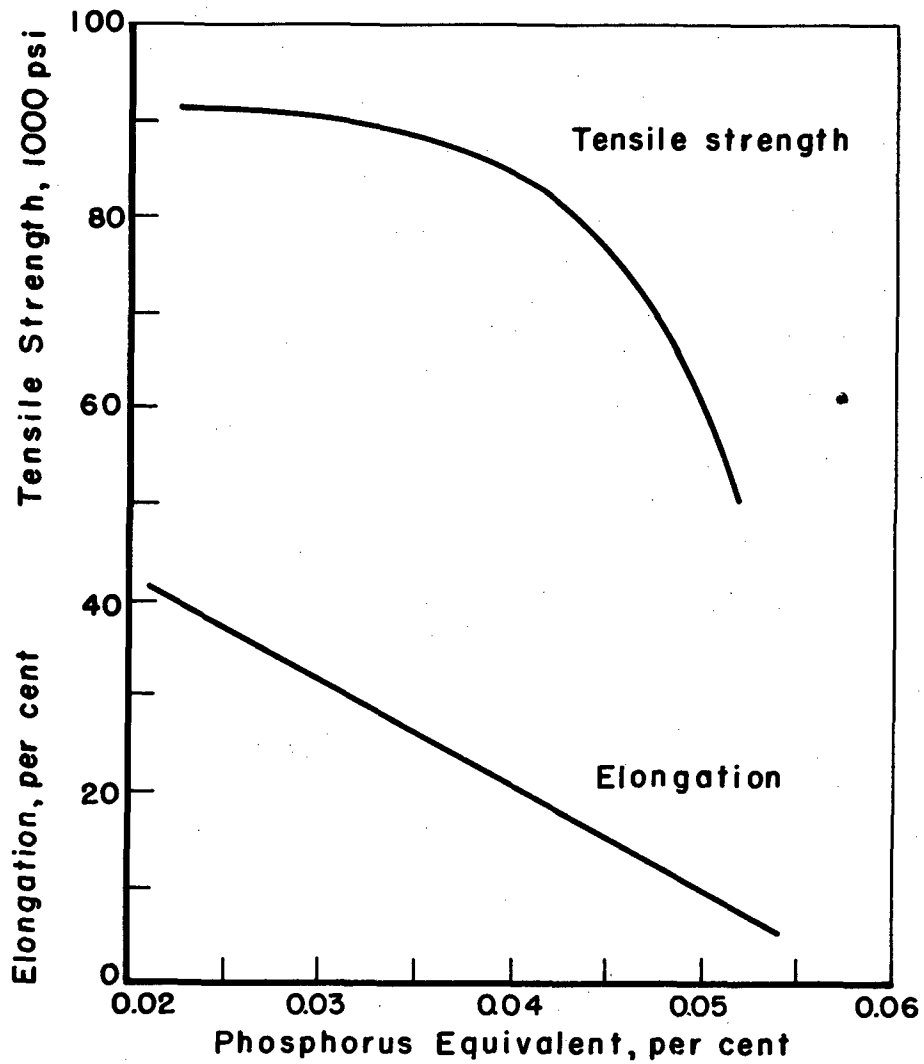


FIGURE 37. EFFECT OF PHOSPHORUS EQUIVALENT ON TENSILE STRENGTH AND ELONGATION OF TYPE 310 WELD METAL<sup>(I3)</sup>

The ductility of Type 310 (25 Cr - 20 Ni) weld metal decreased with increase in sulfur<sup>(13)</sup>. Specimens with low ductility at room temperature showed numerous fine fissures characteristic of hot-short metal. Sulfur in amounts exceeding 0.025 per cent also caused weld-bead cracks and fissuring in weld metal deposited with Type 330 (15 Cr - 35 Ni) electrodes<sup>(134)</sup>. Sulfur and phosphorus have similar effects on Type 330 weld metal and therefore both were plotted in the graph shown in Figure 38 for comparison.

Carbon. Cracking in austenitic weld metal has sometimes been blamed on carbon because carbide precipitation was observed in the vicinity of fine fissures in the weld metal<sup>(145)</sup>. However, the evidence seems to be quite to the contrary, because very low-carbon austenitic weld metal is more susceptible to hot cracking than that containing the normal amount of carbon. It may be that some carbon is necessary to tie up harmful elements as carbides (e. g., columbium).

Carbon increases the tensile strength of austenitic weld metal. It also increases the ductility of the weld metal, up to an optimum value above which the ductility gradually decreases with carbon<sup>(123)</sup>.

Thomas<sup>(13)</sup> found that carbon-strengthened weld metal deposited with both lime-type and titania-type coated Type 310 (25 Cr - 20 Ni) electrodes, and also caused a corresponding decrease in ductility above about 0.10 per cent carbon. Below 0.10 per cent carbon, the ductility apparently decreases with a decrease in carbon, as shown by the curves of elongation versus carbon content in Figure 39. The results of tensile tests made by Linnert and Bloom<sup>(124)</sup> on Type 310 (25 Cr - 20 Ni) weld-metal specimens agreed very well with those conducted by Thomas<sup>(134)</sup>. They stated that the sharp drop in elongation below approximately 0.12 per cent carbon was due to the presence of intergranular microfissures in the weld metal. Carpenter and Jessen<sup>(15)</sup> concluded that the preferred carbon range for Type 310 weld metal was 0.12 - 0.20 per cent for maximum ductility.

The tensile strength of Type 330 (15 Cr - 35 Ni) weld metal was found to increase with increase in carbon<sup>(134)</sup>. Ductility increased with carbon percentages up to about 0.20 per cent carbon and then decreased with further increase in carbon content. The effect of carbon on the elongation of Type 330 is also shown in Figure 39. Fissures were observed in the elongated tensile specimens and, in extreme cases, weld-bead cracks were encountered when the carbon content was below the desired percentage of maximum ductility. In high-silicon (1.5 per cent) electrodes used for welding certain alloy castings, the carbon must be increased to about 0.40 per cent to avoid hot cracks and fissures.

Lee<sup>(120)</sup> made a statistical study of the bead-cracking susceptibility of weld metal deposited with Type 307 (19 Cr - 9 Ni-Mn) and Type 308 (20 Cr - 10 Ni) electrodes. He used a "cracking index" for each heat of core wire, that is, an arbitrary number derived from the amount of extra chromium

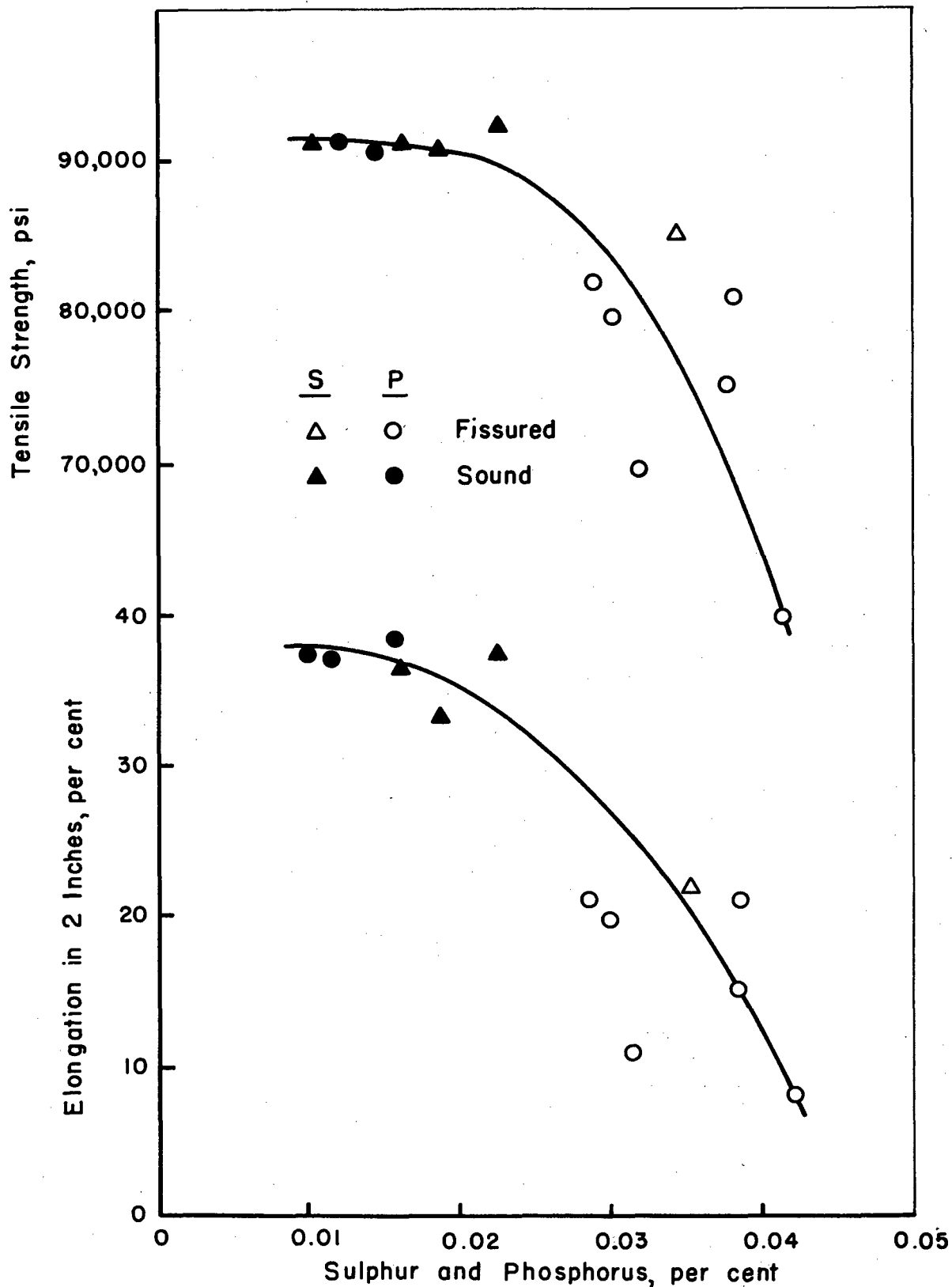


FIGURE 38. EFFECT OF SULFUR AND PHOSPHORUS ON TENSILE STRENGTH AND ELONGATION OF TYPE 330 WELD METAL (I34)

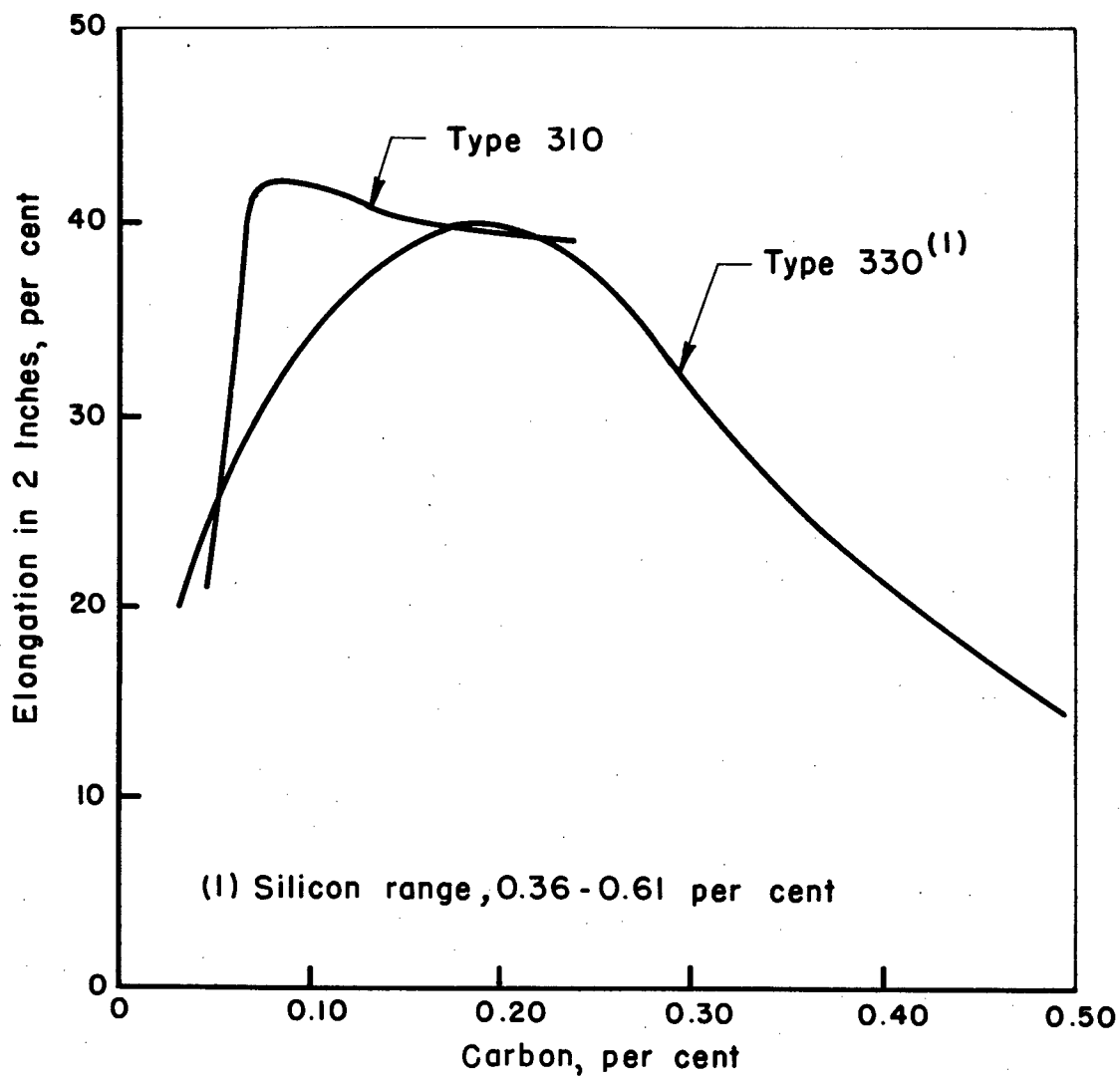


FIGURE 39. EFFECT OF CARBON ON ELONGATION OF ALL-WELD-METAL TENSILE SPECIMENS (I 3, I 34)

which had to be added to the core wire to prevent root bead cracking. The results of tests with core wire with approximately constant chromium-to-nickel ratio and varying carbon content are shown in Table 9. Note that as the carbon increased the phosphorus content also increased. This was apparently a function of steelmaking practice. The authors concluded that cracking susceptibility depended upon the combined effects of carbon and phosphorus.

Carpenter and Jessen<sup>(16)</sup> observed fissuring in Type 347 (19 Cr - 9 Ni-Cb) manual-arc weld deposits with a low carbon content (below 0.10 per cent). The cracking occurred even though the silicon content was kept on the low side (0.39 per cent). An increase in carbon to 0.18 per cent completely eliminated all fissuring and substantially improved both the tensile strength and elongation, even with silicon on the high side (0.70 per cent).

Nitrogen. Nitrogen was found to have a beneficial effect on the yield strength and ductility of cast austenitic steels at 1800°F. For optimum properties, a nitrogen content of about 0.08 to 0.10 per cent was recommended for 23 to 27 per cent chromium - 10 to 13 per cent nickel steels<sup>(112)</sup>. A producer of electrode core wire found that the hot workability of 25 Cr - 12 Ni steel was improved by adding nitrogen<sup>(120)</sup>. However, weld metal deposited with the same core wire cracked very badly.

Nickel. Nickel is used in stainless steel for increased resistance to nonoxidizing corrosives. In austenitic weld metal, nickel tends to increase susceptibility to cracking, not only because it tends to suppress delta-ferrite formation, but also because it appears to increase cracking in fully austenitic structures. The effects of nickel are controlled largely by proper adjustment of the chromium-nickel ratio.

Chromium. The remarkable resistance of stainless steels to oxidizing corrosives is due mostly to the chromium content. Chromium is a ferrite former and nickel is an austenite former, so by properly balancing these two elements, the desired microstructure can be obtained in austenitic weld metal. As previously mentioned, it has been shown by several investigators that the presence of at least 5 per cent delta ferrite in the weld metal has a favorable effect on reducing cracking in austenitic welds.

Chromium is estimated to be about one-half as effective as columbium and about two-thirds as effective as molybdenum in promoting the formation of delta ferrite. Campbell and Thomas<sup>(13)</sup> used the following formula in calculating the chromium equivalent in austenitic weld metal:

$$\text{Chromium equivalent} = \text{Cr} + 1.5 \text{ Mo} + 2 \text{ Cb.}$$

TABLE 9. RESULTS OF TESTS TO SHOW THE EFFECT OF CARBON ON CRACKING IN AUSTENITIC WELD METAL WITH SIMILAR CHROMIUM-NICKEL RATIO<sup>(120)</sup>

Group Carbon, per cent	Number of Heats	Chemical Composition, %				Chromium- Nickel Ratio	Cracking Index
		P	S	Cr	Ni		
0.05-0.07	16	0.017	0.011	20.7	9.8	2.12	65
0.08-0.09	29	0.019	0.011	20.5	9.7	2.12	67
0.10	16	0.020	0.013	20.7	9.8	2.12	69
0.11	19	0.021	0.013	20.9	9.6	2.13	75
0.12	42	0.020	0.011	20.8	9.9	2.09	98
0.13	28	0.021	0.012	20.7	9.8	2.10	90
0.14-0.15	21	0.023	0.013	20.6	9.7	2.12	135

Since chromium promotes delta ferrite, the cracking tendency of austenitic (20 Cr - 10 Ni) steel weld metal was found to decrease with increase in chromium-nickel ratio<sup>(18)</sup>. A minimum ratio of 2 to 1 with about 0.75 per cent molybdenum was recommended for austenitic electrodes used in welding armor in order to prevent weld-metal cracks and low root-cracking sensitivity, and to obtain optimum ductility. The minimum chromium-nickel ratio of 2 to 1 is specified by both the Army and the Navy for Type 307 (19 Cr - 9 Ni-Mo) and Type 308 (20 Cr - 10 Ni) electrodes<sup>(120)</sup>.

Copper. Additions of copper in percentages 2 to 14 per cent markedly increased cracking in weld metal deposited with Type 316 (18 Cr - 12 Ni - 3 Mo) electrodes<sup>(151)</sup>. This is very effectively illustrated in Figure 40A for weld metal which contained 13.5 per cent copper. Note that the cracking extended for some distance into the base metal, which was Timken Alloy (16 Cr - 25 Ni - 6 Mo). The free copper in the grain boundaries of the weld metal shown in Figure 40B is supporting evidence that low-melting grain-boundary constituents can cause cracking.

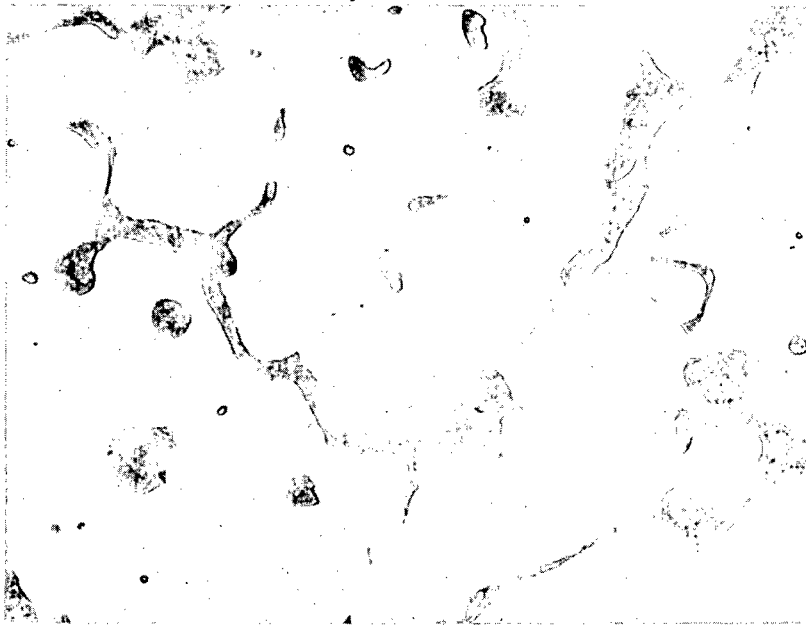




8X

Ferric Chloride Etch

A. Badly cracked weld



1000X

$\text{NH}_4\text{OH} + \text{H}_2\text{O}_2$  Etch

B. Free copper in grain boundaries

FIGURE 40. CRACKS AND GRAIN-BOUNDARY STRUCTURE IN AUSTENITIC WELD METAL CONTAINING 13.5 PER CENT COPPER

## Hot Ductility and Hot Strength

In their studies of the causes of weld-metal cracking, a number of investigators have devised methods for measuring the hot ductility and strength of weld metal. One objection to the tests used so far by most investigators is that measurements were made on reheated samples only, instead of on samples cooling from the solidification temperature.

Goodger<sup>(H104)</sup> described some early hot-shortness tests on welds. One of these consisted of two semicircular plates which were welded together to form a circular plate about 12 inches in diameter. This plate was then heated to a temperature of 1560°F to 1650°F and pressed through a die to a cup form. This was similar to the Ericssen test for sheet metal. The results of a number of tests showed that weld metal deposited by the oxyacetylene process stood up quite satisfactorily, whereas welds made with metal-arc electrodes (bare wire) cracked in all directions when hot formed. Ferritic metal with the lowest carbon content gave the worst results (this indicated that FeO was the cause of the hot shortness). Hot-bend tests made at 1650°F by Goodger<sup>(H104)</sup> also showed that the hot-shortness tendency of weld metal deposited with ferritic electrodes decreased with increase in carbon in the range 0.08 to 1.20 per cent. The author suggested that the lack of ductility in the low-carbon-steel welds was due to ultramicroscopic films of oxide.

The results of hot-tension tests showed that mild-steel weld metal deposited with Class E217 (British) electrodes had minimum ductility in the temperature range 1650°F to 2010°F<sup>(H25)</sup>. The authors did not know whether hot cracking actually occurred within this temperature range, but they did observe that one weld metal (A), particularly susceptible to hot cracking, had a lower minimum ductility than another weld metal (B), which was relatively free of hot-cracking tendencies. The results of hot-tension tests on these weld metals, parallel to the direction of welding, are shown in Figure 41. The compositions of the two weld-metal deposits were:

	Composition, Per Cent				
	<u>C</u>	<u>Si</u>	<u>Mn</u>	<u>S</u>	<u>P</u>
Electrode A	0.14	0.127	0.39	0.061	0.024
Electrode B	0.10	0.122	0.62	0.044	0.040

It was also found that the low ductility was more pronounced on specimens taken transverse to the direction of welding. The effect of iron oxide and sulfur on hot ductility was studied but the results were not conclusive. This would indicate that the low-ductility region was probably not related to hot cracking in weld metal.

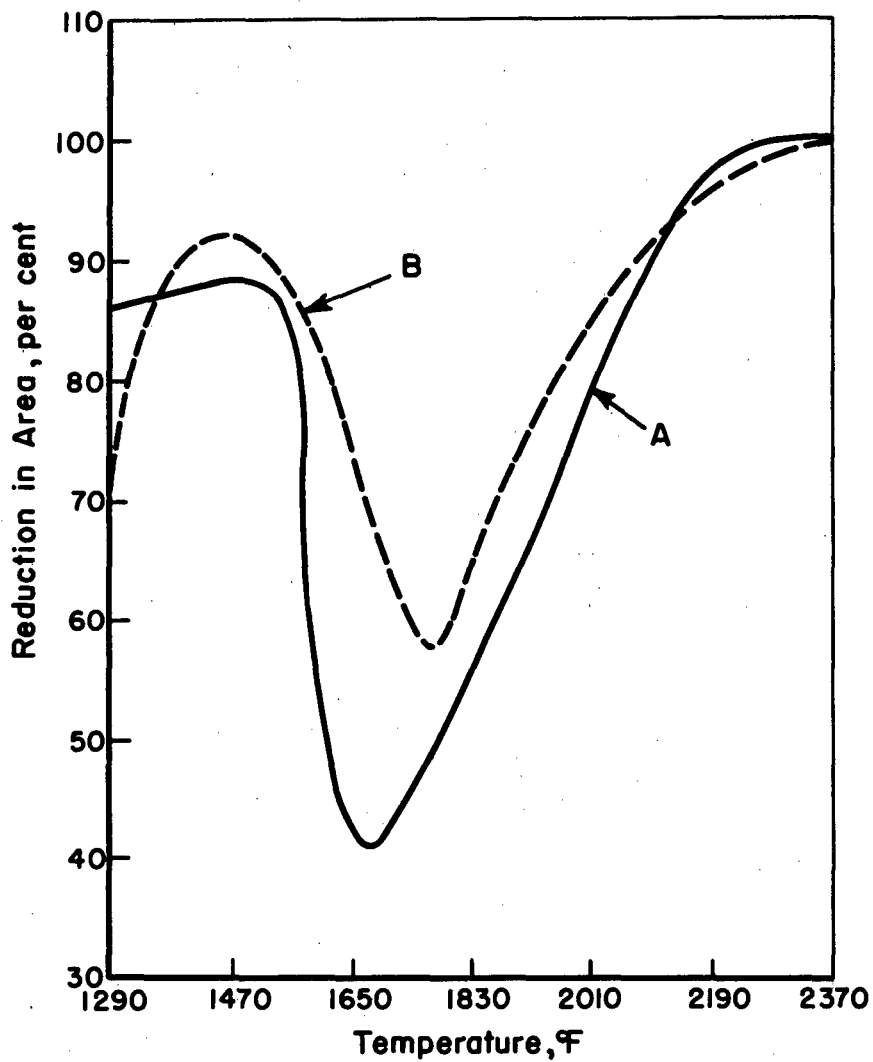


FIGURE 41. HIGH-TEMPERATURE DUCTILITY OF TWO WELD-METAL COMPOSITIONS: (A) SUSCEPTIBLE TO HOT CRACKING, AND (B) RELATIVELY CRACK FREE <sup>(H25)</sup>

Rollason<sup>(H211)</sup> made tension tests of core wire and weld-metal samples in the temperature range 2190°F to 2550°F, using rapid resistance heating and a rapid rate of loading. He observed a drop in ductility, for both the core wires and the weld metals, in the temperature range 2280°F to 2335°F, as shown in Figure 42. The core wires and deposited weld metal had the following composition:

Type	Composition, Per Cent				
	C	Mn	Si	S	P
F R Wire	0.13	0.39	0.01	0.034	0.049
F W Weld	0.12	0.52	0.15	0.022	0.072
R R Wire	0.11	0.40	0.015	0.041	0.024
R W Weld	0.10	0.53	0.16	0.028	0.042

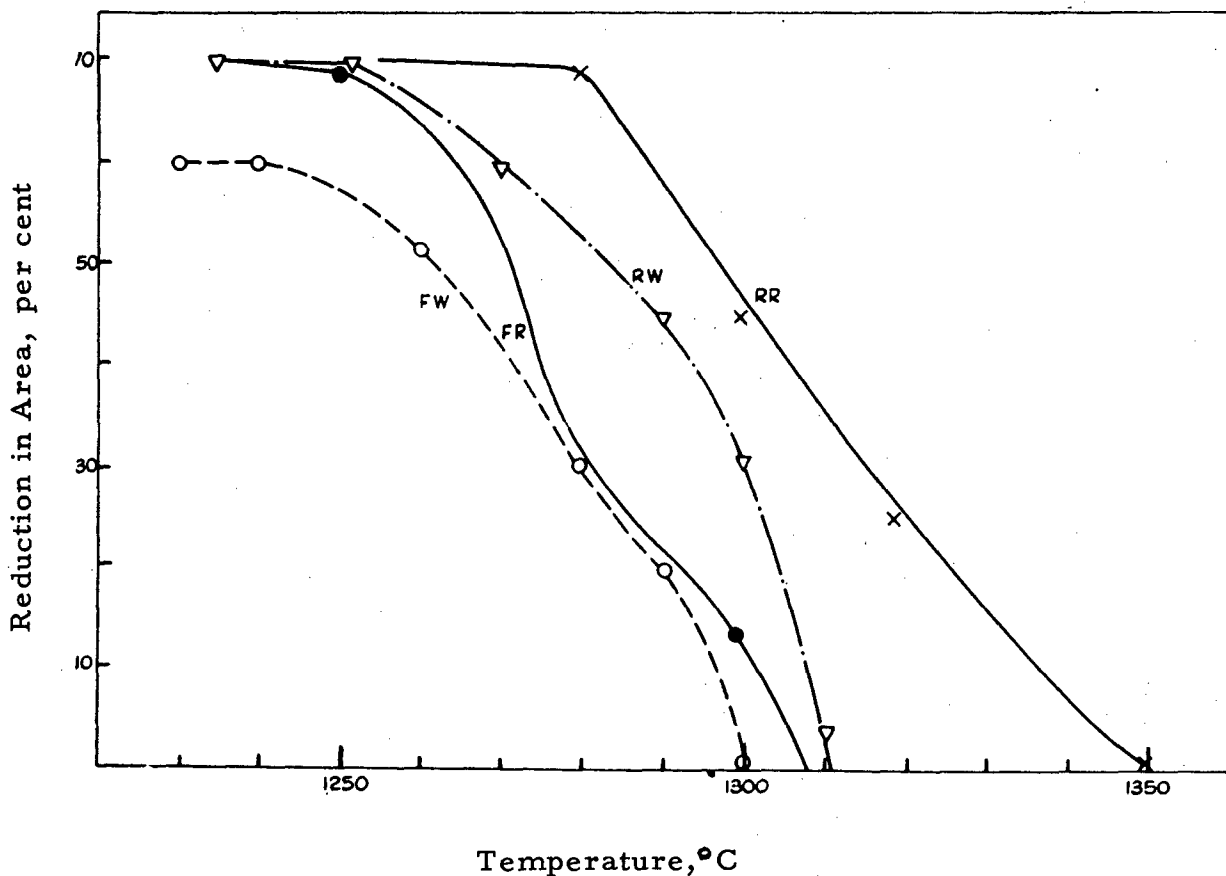


FIGURE 42. CURVES SHOWING THE HIGH-TEMPERATURE DUCTILITY OF CORE WIRES AND THEIR RESPECTIVE WELD-METAL DEPOSITS(H211)

Weld metal deposited with Type F electrodes was highly susceptible to cracking in the Murex Hot-Cracking Machine, whereas the Type R weld metal was crack resistant. This difference in cracking susceptibility was reflected in the high-temperature ductility measurements, since

the FW weld metal and FR core wire showed a drop in ductility at a lower temperature than the RW weld metal and RR core wire. It also should be noted that the core wires had slightly higher ductility than their respective weld metal.

High-temperature ductility studies were made at Battelle on a series of gas-turbine alloys and weld metals, in connection with an investigation on the causes of weld-metal cracking in heat-resisting alloys<sup>(151)</sup>. Tension tests were made in the temperature range 2100°F to 2400°F. For each alloy and weld metal, the approximate temperature was determined at which the character of the tension test-fracture changed from ductile to nonductile. The weld-metal specimens were tested transverse to the direction of welding. The lowest temperature at which the material exhibited little or no ductility was selected as the hot-shortness temperature of these alloys, as listed below:

Material	Approximate Hot-Shortness Temperature, °F
<u>Unwelded Specimens (1)</u>	
Timken Alloy	2325
Vitallium	2175
S-816	2300
Inconel "X"	2275
Type 316 electrode core wire	2475
Type 349 electrode core wire	2425
Type 310 electrode core wire	2400

Material	Approximate Hot-Shortness Temperature, °F
<u>Weld-Joint Specimens</u>	
Type 316 electrode - Timken alloy	2325
Type 349 electrode - Timken alloy	2375
Type 316 electrode - Vitallium	2275
Type 349 electrode - Vitallium	Above 2250

(1) The specimens and equipment used in these tests will be described later in this report.

It was interesting that weld metals deposited with the Type 316 and Type 349 electrodes showed a nonductile fracture at a lower temperature than their respective weld-joint specimens. The weld metal deposited with Type 349 had been shown, in circular-groove welding-cracking tests, to

be less crack sensitive than Type 316 weld metal. This agreed somewhat with the hot-shortness tests. Also, it was significant that Type 349 filler wire showed much higher ductility than any of the other materials below its hot-shortness temperature.

Clarke<sup>(H59)</sup> made hot-bend tests on various weld deposits made on high-temperature alloys (no mention of composition). He observed that the welds, in general, broke with little or no strength in the vicinity of 2325°F.

### Blue Brittleness in Weld Metal

Cracking in weld metal which occurs below the austenite-ferrite transformation temperature has sometimes been attributed to the blue-brittleness phenomenon. Blue brittleness in steel was described earlier in this report. Sayre<sup>(H223)</sup> believed that blue brittleness was a factor in weld-metal cracking, but he was unable to demonstrate this.

Hodge<sup>(H125)</sup> observed that the blue brittleness of weld metals deposited with bare and covered electrodes increased with the nitrogen content as determined by torsion tests, impact tests, and tension tests of welds heated in the blue-heat range. Weld metal deposited with covered electrodes had the highest notched-bar impact strength. Blue brittleness is manifested by a drop in impact strength and ductility and increase in tensile strength. The maximum effect of blue brittleness was obtained in the temperature range 805°F to 985°F. This author did not attempt to relate weld-metal cracking to blue brittleness, however.

Eckel and Raudebaugh<sup>(H83)</sup> made a series of elevated-temperature impact tests of weld metal deposited with E6012, E6020, and Type 309 stainless steel electrodes. Results on a series of Charpy keyhole-notch specimens tested in the temperature range 80°F to 1600°F are illustrated by the curves shown in Figure 43. The welds were tested transverse to the direction of welding. The weld metals deposited with E6012 and E6020 electrodes showed minimum impact strengths at about 1000°F. No such minimum was found in welds made with Type 309 stainless electrodes. The authors felt that the low impact strength of the E6012 and E6020 welds at the 1000°F level might be due to blue brittleness and might explain the hot-cracking tendencies of these weld metals. Blue oxide observed on crack surfaces suggested that they did not occur at excessively high temperatures. This investigation also showed that the minimum impact strength of the mild-steel samples could be eliminated by reducing carbon and nitrogen. Recarburizing or renitriding caused a return of the minimum impact strength in these steel samples at 1000°F. The investigators suggested that the presence of a carbide precipitate and possibly a nitride precipitate at the ferrite grain boundaries of weld metal may result in the formation of hot cracks. Blue brittleness has been attributed to carbon and nitrogen.

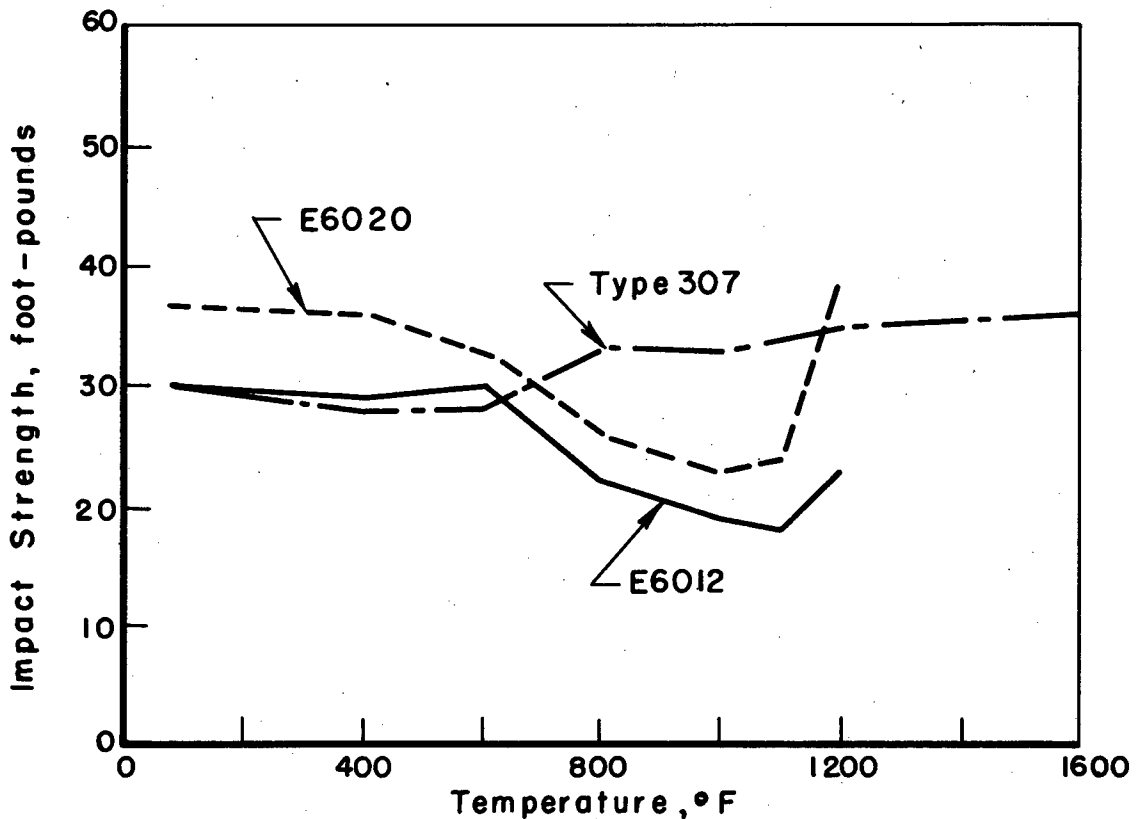


FIGURE 43. EFFECT OF TEMPERATURE ON THE CHARPY KEYHOLE-NOTCH IMPACT STRENGTH OF WELD METAL<sup>(H83)</sup>

Using the Lehigh restraint test, Stout, et al<sup>(H271)</sup>, observed that the cracking temperature (212°-320°F) of weld metal deposited with E6010 electrodes had no detectable relation to the cooling rate or the amount of restraint. They suggested that the cause of weld-metal cracking at these temperatures might be blue brittleness at the time when maximum stress was applied to the weld.

Gaymans<sup>(H102)</sup> stated that the blue-brittleness phenomenon might explain cracking in weld metal deposited with E6012 and E6013 electrodes, which could not otherwise be explained. He reasoned that strain rates in the deposited weld metal might be high enough to raise the blue-brittleness temperature as high as 1100°F. It was pointed out, however, that cracking occurring only below the austenite-ferrite transformation could possibly be related to blue brittleness. Since austenitic weld metal does not transform, it is not subject to blue brittleness.

## Strain Aging in Weld Metal

As described earlier in this report, strain-aging embrittlement is a general term applied to the continued change in physical properties when a cold-worked steel is allowed to stand at room temperature or is heated at a low temperature for a short time. A number of investigators have studied strain aging in weld metal. Weld metal deposited with bare-wire electrodes was shown to be susceptible to strain-aging effects, believed to be caused by nitrogen pickup during welding<sup>(H80, H118, H132, H205)</sup>. Nitrogen was also found to cause strain-aging embrittlement in coated-electrode deposits<sup>(H80, H162)</sup>. Various investigators<sup>(H47, H48, H96, H98)</sup> have used notched-bend tests and Charpy vee-notch tests to study strain-aging behavior of AWS Classes E6010, E6012, E6020, and E6016 electrodes in connection with transition-temperature studies.

Since it is an embrittlement phenomenon, strain aging was included in this survey. However, in no instance did any investigator suggest that it was related to weld-metal cracking. Therefore, the discussion on the effects of strain aging was kept to a minimum.

### Temperature at Which Weld-Metal Cracking Occurs

The mechanism of hot cracking might be better understood if the temperature of its occurrence were known. Most investigators agree that hot cracking in ferritic weld metal occurs at very high temperatures, but there is some question as to whether it initiates above or below the solidus temperature. Cracking in austenitic weld metal was generally believed to occur very soon after solidification of the deposit. So far, apparently, no one has succeeded in measuring the actual temperature of weld-metal cracking.

Several German investigators have studied the problem of base-metal hot cracking encountered in the gas welding of thin aircraft steel sheet. Bardenheuer and Bottenberg<sup>(H26)</sup> attached thermocouple wires 3 mm from the weld and measured temperatures in the range 1110°F to 1365°F when cracks were first observed. The authors assumed that the temperatures were too low. They were probably correct in this assumption, since it has been shown that surface temperatures cannot be accurately measured with exposed thermocouples. Muller<sup>(H184)</sup> made similar measurements and concluded that the cracks developed at least above 1830°F.

Antonioli<sup>(H18)</sup> used a rather unique setup to measure the temperature at which the same type of base-metal hot cracking occurred. A schematic diagram of his setup is shown in Figure 44.



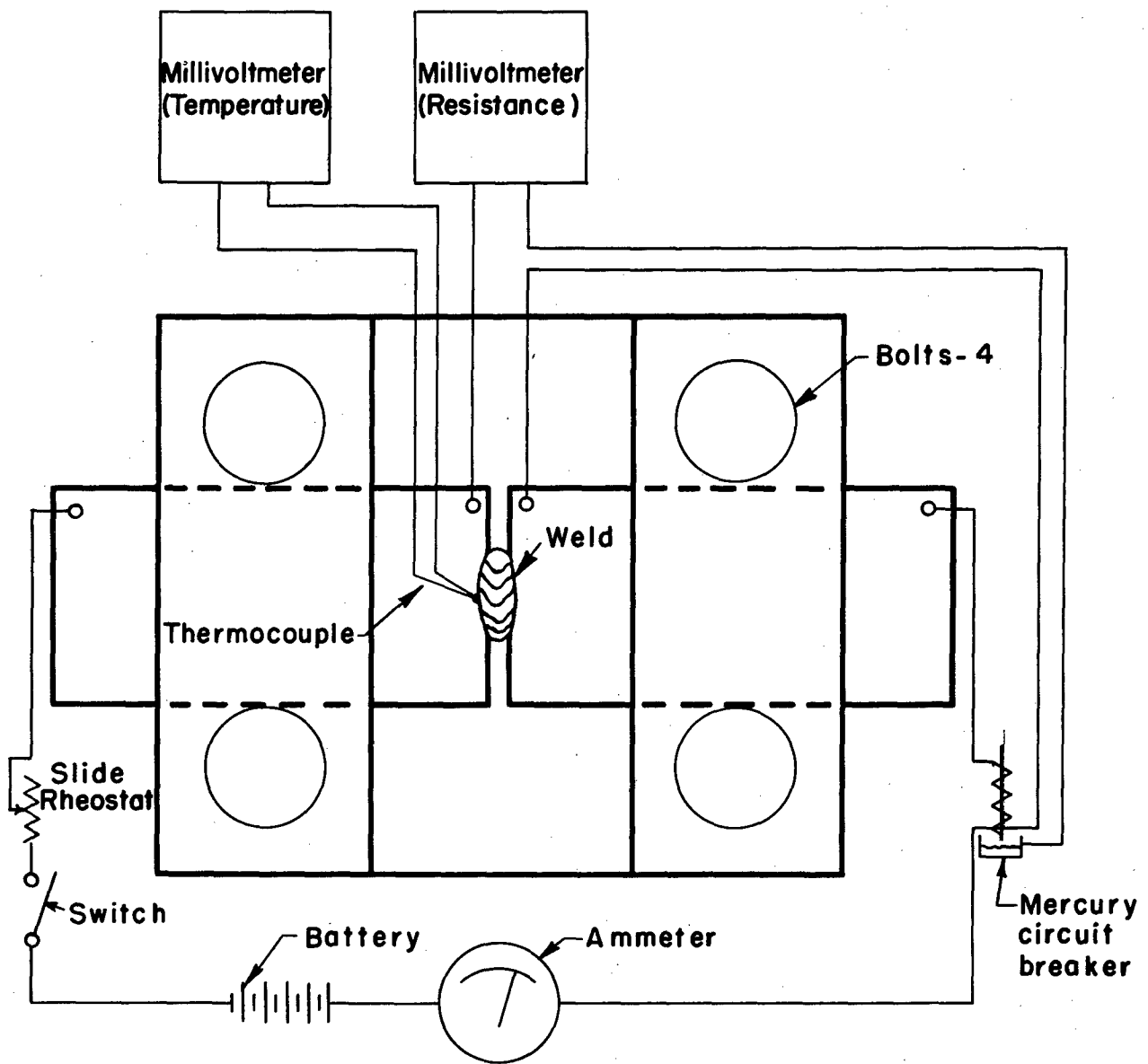


FIGURE 44. APPARATUS FOR MEASURING THE TEMPERATURE OF CRACK FORMATION IN GAS-WELDED STEEL PLATE<sup>(H18)</sup>

Using this temperature-measuring setup, he concluded that base-metal hot cracking in chromium-molybdenum steel (0.26-0.30 C, 0.36 Si, 0.52-0.73 Mn, 0.83-1.13 Cr, 0.03 P, 0.07 S) sheet occurred in the temperature range 2460°F to 1830°F, with the majority occurring at 2320°F.

Helin and Svantesson<sup>(H116)</sup> observed cracking with ferritic electrodes (no mention of composition) to occur almost immediately after the slag had come to rest on the weld metal. They also kept welded joints at white to red heat during welding, and showed that hot cracking occurred at temperatures well above 1770°F. Rollason<sup>(H215)</sup> observed cracks in mild-steel weld metal occurring while the arc was still playing on the crater.

In tests with Lehigh restraint specimens, cracks were found to form at about 395°F in weld beads deposited with E6010 electrodes<sup>(H271)</sup>. These were not referred to as hot cracks or cold cracks. It was suggested that the cause of cracking was possibly blue brittleness at the time when maximum stress was applied to the weld.

Hess<sup>(H123)</sup> did not believe that cracking in ferritic welds of a given composition occurred at specific temperatures. He said that the temperature of cracking depends upon the stress condition at the time.

In an investigation by Steinberger, et al<sup>(H264)</sup>, cracks appeared in welds deposited with Type 410 martensitic stainless steel electrodes, except when a 600°F postheat temperature was used. This temperature was critical because cracking occurred at both higher and lower temperatures with this electrode. In a discussion of the paper, Bland<sup>(H38)</sup> postulated that cracking below 600°F postheat was due to hydrogen, whereas cracking above this temperature might be due primarily to stresses produced by structural changes alone.

Hot cracking in austenitic stainless steel weld metal is believed by most investigators to initiate at very high temperatures, perhaps above 2300°F, and is due to hot tearing or hot shortness. Gayley<sup>(I11)</sup> has observed that hot cracks in welds made with Types 307, 308, and 310 steel electrodes were intergranular and at right angles to the isothermal lines. Thielsch<sup>(I40)</sup> mentioned that slag was usually found in cracks in submerged-melt welds made with 18 Cr-13 Ni-1 Cb electrodes, which indicated that they opened up at very high temperatures when the slag was still molten. Similarly, other investigators<sup>(I51)</sup> observed slag in welds made with Type 316 (18 Cr-12 Ni-3 Mo). In melting-point determinations of the type of slag found in the cracks, the slag softened at 2350°F and finally melted at 2450°F. This indicated that the cracking had occurred at, or above, 2350°F to 2450°F.

Some investigators<sup>(I7, I22)</sup> attributed hot cracking in austenitic weld metal to low-melting interdendritic segregates which they observed in metallographic studies. The segregates were believed to remain fluid for

a longer period of time than normally, causing a weak zone which could not withstand the stresses built up during solidification.

As previously discussed, the susceptibility of austenitic weld metal to cracking is attributed to its greater propensity to hot shortness<sup>(I5, I7, I17, I45)</sup>. Since the recommended forging range for austenitic steels is 1700°F to 2200°F, the austenitic welds would be expected to be hot short only at temperatures above 2200°F.

## METHODS USED TO CONTROL WELD-METAL CRACKING

In attempting to avoid weld-metal cracking, many investigators and welding engineers have used special welding procedures. Most of these were designed to reduce the stress across the weld joint, and some involved metallurgical changes in the deposited weld metal.

### Welding Technique

The human element is a very important factor in controlling weld-metal cracking. The statement has been made that "a highly skilled welder can sometimes make a sound deposit with an electrode known to be susceptible to hot cracking by unconsciously repairing minute cracks appearing in the semimolten pool"<sup>(H215)</sup>. Helin and Svantesson<sup>(H116)</sup> observed that a backward and forward movement during welding often covered or eradicated weld-metal cracks as fast as they were formed. Other investigators<sup>(H164)</sup> reported that cracks in welds with narrow beads were eliminated in the same material by using a weaving bead.

In Germany, a change in welding technique reduced weld-metal cracking in joining chromium-molybdenum aircraft steel tubes using the oxyacetylene welding process<sup>(H41)</sup>. A backhand technique produced much less cracking than a forehand technique. On the other hand, another German investigator<sup>(H185)</sup> described difficulties with base-metal hot cracking in oxyacetylene welding of chromium-molybdenum steel tubes. He found that welding techniques appeared to have no effect on the cracking, and concluded that the fundamental cause lay in the composition of the material, particularly in the phosphorus and sulfur contents.

For years, welders have been using stepback, block, and cascade sequence techniques to prevent distortion and cracking in especially difficult joints. These techniques have been more or less adopted by welders as part of the art and "know how" of welding. Very little information on these techniques was found in the literature. However, they are amply described in the Welding Handbook, Third Edition.

## Prestressing

Application of a compressive stress on a joint during welding has been successfully used in some instances to prevent weld-metal cracking. The primary function of compressive stress was to offset the tension stresses developed during cooling. Hot cracking in weld-metal deposits on mild-steel structures was minimized by using jacks to maintain the compressive stress during welding<sup>(H44, H155)</sup>. The method was not practical, however, and a simple change in welding sequence proved to be just as effective.

Circular patches for filling up holes (such as portholes in ship hulls) are very difficult to weld without weld-metal cracking because of the severe restraint imposed. Transverse stresses of 40,000 to 50,000 psi have been measured in such a joint<sup>(H123)</sup>. One method used to prevent this cracking was to apply localized heating on one spot on the base metal near the start and finish of the weld before welding<sup>(H123)</sup>. The heating supplied sufficient residual compressive stress in the base metal to prevent cracking. Another method used successfully was to drive a wedge in the joint before welding, weld around to the wedge and then knock it out before completing the weld. Compressive stresses due to the wedge reduced the strain on the weld crater. A third method was to dish the patches prior to welding. In filling up portholes, such dished patches were actually flattened out during welding because of the high transverse stresses.

Shrink fitting shafts through plates or flywheels, for example, was a means of acquiring the desired compressive stress to prevent weld-metal cracking at the joint<sup>(H164)</sup>. A similar method was used to reduce cracking in flange welds on aircraft steel tubes<sup>(H107)</sup>. The flanges were heated to 500°F on the inside edge and to 800°F to 1000°F on the outside edge before placing them on the tubes. The flanges were then allowed to cool on the tubes so that they exerted a compressive stress. During welding, the inside of the tube was heated to maintain the stress.

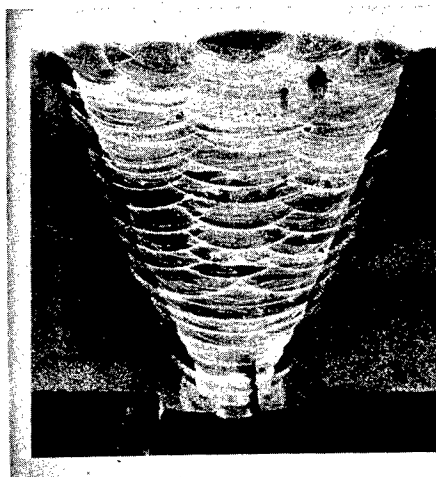
Mechanical predistortion was used to reduce transverse stress which caused cracking in seam welds on SAE 4130 tubing<sup>(H221)</sup>. The residual stress set up by the predistortion was actually stress relieved during oxy-acetylene welding of the tubing.

## Peening

Peening is the mechanical working of metals by hammer blows. In welding, it is mainly a stress-relief treatment, and it has been suggested that it might reduce weld-metal cracking. Peening tends to refine the grain of ferritic welds and increases their strength<sup>(H184)</sup>. These effects

are very shallow, however. Peening is most effective when the weld is at a dull red heat. There is the danger, however, that cracking will result if the weld is peened at some critical temperature.

Peening is widely used as a means of preventing distortion in multipass weld deposits in heavy plate. Each weld pass is peened before laying the subsequent pass. This stress relieves the weld deposits and prevents the build-up of stress which causes base-metal cracking failure, which frequently occurs well away from the weld<sup>(H51, H164)</sup>. Peening also prevents root-bead failures in multipass welds in heavy plate such as that shown in Figure 45<sup>(H101)</sup>. Without the benefit of peening, the stresses in such a joint were known to exceed 60,000 psi.



1X

FIGURE 45. CRACK IN ROOT OF WELD IN HEAVY PLATE CAUSED BY STRESS EXCEEDING FRACTURE STRENGTH<sup>(H101)</sup>

In multipass austenitic welds, peening was used to break up the columnar grain structure which lowers the tensile strength of the weld metal<sup>(I11)</sup>. Its function was to cold work the weld bead, which was then recrystallized by the heat applied during the deposition of subsequent layers.

Transverse cracking was a major problem in welding pressure vessels fabricated from 3- $\frac{1}{2}$ -inch-thick SAE 6125 (chromium-vanadium) steel plate<sup>(H164)</sup>. The vessels had to be welded from the outside using a deep U-groove joint. Longitudinal cracking was eliminated by a change in joint design, but the transverse cracking persisted. The transverse cracking was finally prevented by peening each layer of the weld and also by thermal stress relieving after each one-inch-thick layer of weld was deposited.

Although peening is a useful tool in post-weld stress relief, there appears to be little evidence that it has much effect on hot cracking in

weld metal. One reason for this is that hot cracking probably occurs long before the peening tool can be applied.

Several examples of employing peening to prevent cracking in hard-facing deposits were reported. These cracks were believed to be due to the differences in dilation characteristics between the base metal and the facing deposit. Hall<sup>(H110)</sup> prevented check cracking or surface cracking in high-manganese (14 per cent) steel hard-facing deposits by peening. The effect of hammering during cooling was to expand the deposit and remove all contraction strains. Magee<sup>(H176)</sup> also prevented cracking in Hadfield steel hard-facing deposits by peening. Small pads were applied (one or two square inches) and then immediately quenched with cold water and peened while still very hot. Peening broke the light slag coating and stress relieved the weld deposit underneath. Avery<sup>(H22)</sup> also recommended peening for relief of residual tensile stresses of hard-facing deposits but only for those of the ductile austenitic type.

### Preheat

The value of preheat in preventing weld-metal cracking varies considerably with different weld metals. Preheat may have two separate functions which could be opposite in their effects on hot cracking. One effect is the relief of stress on the welded joint, which should reduce hot cracking. Another effect is that of reducing the cooling rate through the hot-cracking temperature range (not established) which is believed to increase cracking. However, preheat is beneficial in preventing cold cracking. Examples of the use of preheat to reduce weld-metal cracking were presented earlier in this report, when discussing the effect of cooling rate.

### Dissimilar Electrodes

Austenitic electrodes are widely used for welding high-strength ferritic steels, particularly armor. Originally, they may have been adopted because of the problem of under-bead cracking. The austenitic weld metal acts as a reservoir to hold hydrogen in solution where it does no harm<sup>(H238)</sup>. Austenitic electrodes are still used to a considerable extent, even with the advent of low-hydrogen electrodes, because of the higher ductility of austenitic weld deposits in service. However, interface cracking has been a problem in austenitic welds on armor. As previously described, this cracking occurs in a narrow zone of low ductility near the fusion line produced by base-metal dilution of the weld.

At least one investigator<sup>(H210)</sup> suggested that welding high-tensile steels with austenitic electrodes reduced weld-metal cracking. This is

difficult to understand when it is considered that austenitic weld metal is notoriously susceptible to hot cracking. Weld-metal cracking was a major problem in welding armor with austenitic electrodes during the Second World War<sup>(H52)</sup>. However, weld-metal cracking was a lesser problem than under-bead cracking which the austenitic electrodes prevented.

### Buttering

In general, when a cracked weld is ground or chipped out and re-deposited, the subsequent weld is less liable to crack<sup>(I2)</sup>. There are, of course, several factors involved, but probably the most important is the precoating effect of the first weld, which had not been entirely removed. High-tensile ferritic electrodes generally have a lower carbon content than the plate being welded. Therefore, the thin layer of weld metal still remaining on the surface of the base plate provides a low ductility material to weld to. Also, the layer of weld metal lessens the dilution effect. The tendency to weld-metal cracking is reduced by both of these factors.

"Buttering" is another means of accomplishing the same end. By this method, a thin deposit of weld metal is applied on the scarf face on each plate before welding. In welding chromium-nickel-molybdenum steels, weld-metal cracking was prevented by first lining the plates with Type 310 (25 Cr-20 Ni) electrode and then filling in with Type 304 (18 Cr-8 Ni) electrode in a single-vee joint<sup>(I13)</sup>. This technique eliminated two types of cracking: interface and hot cracking. Interface cracking was not found with Type 310 electrodes because of their highly austenitic nature, even with some dilution by the base metal. Type 304 weld metal was susceptible to interface cracking but did not hot crack as readily as the Type 310 weld metal. Low-carbon ferritic linings were also used with a high-tensile steel filling with limited success. They were successful if the weld beads were laid in rapid succession; otherwise, cracking resulted.

In some instances, where cracking was a problem in hard facing, it was found advantageous to first butter with a layer of mild steel<sup>(H260)</sup>. This reduced stress due to differences in thermal expansion between the base metal and hard-facing material.

### Eliminating The Weld Crater

The crater of a weld is generally the place most subject to cracking because it is thin in cross section and is the last portion of the weld to solidify. In an effort to prevent cracking at the crater, various methods have been used to eliminate it entirely. One of these is to use run-off tabs, but this is now always convenient in practice. Also, this will not eliminate

the craters in long passes where the welder has to stop and start several times. In such cases, it has proved practicable to have the welders work in pairs<sup>(H116, H243)</sup>. As the one welder is about to finish his electrode, the other welder steals the arc from him and continues the welding, so that no craters are formed. Automatic welding has a similar advantage of continuous welding.

A third method, which has been widely used, is to fill in the craters by thickening the deposit with an oval motion of the electrode making a wide convex surface. This method prevented crater cracking, but had the disadvantage that root cracking often occurred beneath the crater<sup>(I11)</sup>. This was a more serious defect and less likely to be detected. Filling in craters was also bad, because it caused porosity and shrinkage cavities<sup>(H243)</sup>. One investigator<sup>(I11)</sup> had considerable success with a slightly modified technique; however, this was in welding austenitic steels. He prevented cracking by filling the crater and then drawing the arc forward along one side of the scarf. The arc was broken about  $\frac{1}{2}$  inch away from the nominal end of the weld deposit. Another investigator<sup>(I26)</sup> prevented crater cracking in austenitic welds by first filling in the crater and then breaking the arc very slowly. In both of these methods, the craters were filled but humped up above the normal surface of the weld.

#### METHODS USED TO STUDY WELD-METAL CRACKING

The selection of a suitable weld-metal-cracking test is usually guided by the particular conditions encountered in practice. The fact that welding conditions are so varied may partly explain why so many different weld-cracking tests have been developed.

The following factors should be taken into consideration in the selection of a weld-metal-cracking test:

1. Reproducibility of results.
2. Freedom from variation due to the human element.
3. Sensitivity to small changes in a test variable.
4. Ability to show effects of several welding variables.

The weld-metal-cracking tests described in the literature are of four general types: (1) fillet-weld tests, (2) groove-weld tests, (3) controlled-restraint tests, and (4) tests employing externally applied stress. The fillet-weld tests include single-tee joints, double-tee joints, and lap joints. The groove-weld test specimen may have an open root with or without a backup or may simply be a groove in a plate. Controlled-restraint tests such as the Lehigh test, described later, are designed so that the total shrinkage stresses can be varied as a condition of the test. The last group consists of cracking tests in which a tension or bending force is applied to a joint while the weld metal is being deposited.



### Fillet-Welded Tee Joint

The fillet-welded tee-joint test shown in Figure 46 is probably one of the oldest and most commonly used tests for determining the cracking susceptibility of weld metal. It is inexpensive and requires very little joint preparation. Also, the dimensions of the specimen, i. e., plate thickness, can be varied in accordance with the particular conditions encountered in practice.

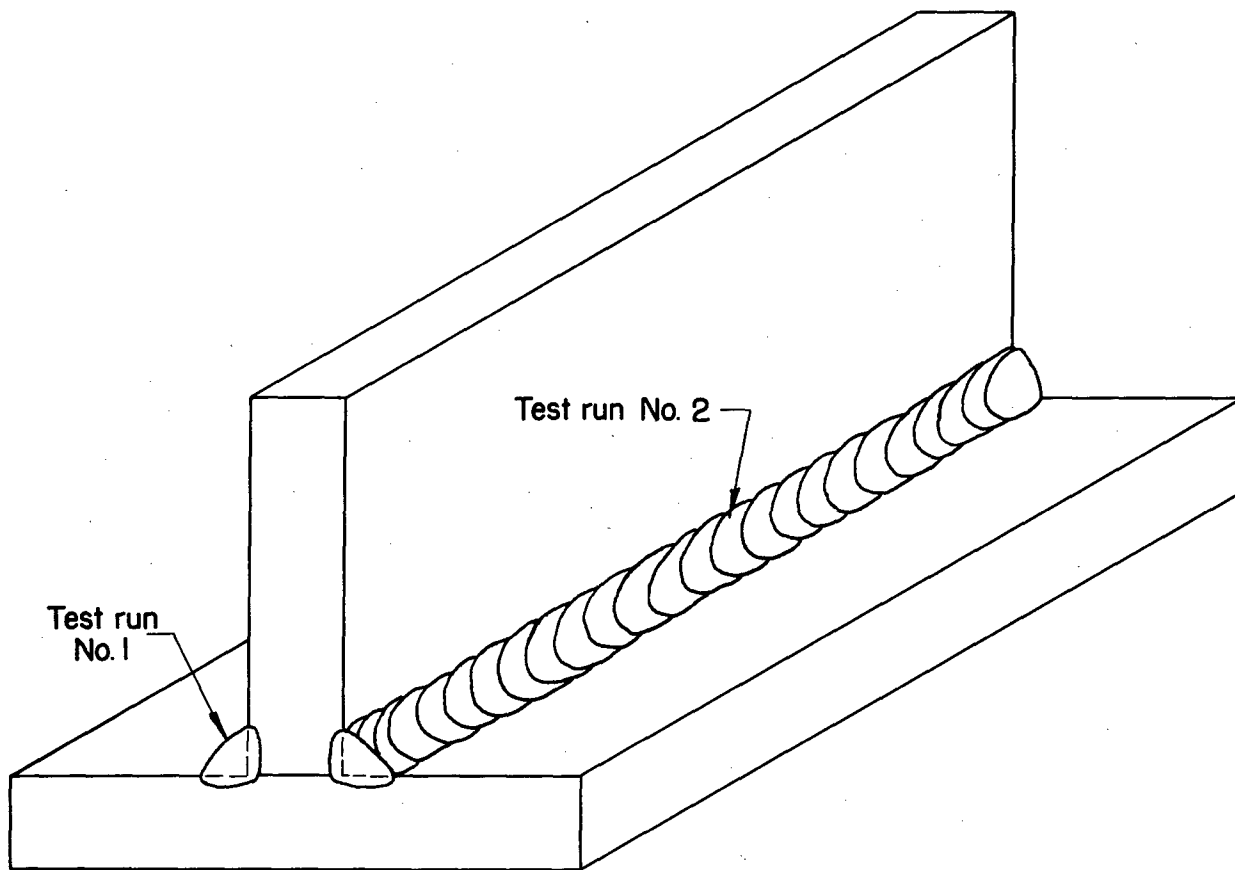


FIGURE 46. SPECIMEN FOR FILLET-WELDED TEE-JOINT TEST

It is recommended that the abutting edge of the vertical plate and the surface of the horizontal plate be finish machined for best results. Rough flame-cut edges tend to reduce the amount of restraint on the fillet welds.

It is reported that highly susceptible welds usually crack on the first run, with the less sensitive cracking on the second run because of the additional restraint afforded by the first weld. The second run is made as

soon as possible after the first run has been completed. Welds which do not crack on either run are rated as crack free. The single-tee fillet test is ideal for quick qualitative studies of various weld-metal compositions<sup>(H61)</sup>.

#### Fillet-Welded Double-Tee Joint

The fillet-welded double-tee-joint test is essentially an advanced modification of the fillet-welded tee-joint test. The test involves the accurate fitup of two vertical members and one horizontal member, as shown in Figure 47. Four fillet welds are then deposited. The severity of the test is increased by cooling the specimen to the original plate temperature before depositing subsequent welds. Each successive weld experiences greater restraint than the previous one. Therefore, weld deposits which crack earliest in the test are most susceptible to cracking. The extent of cracking is determined by magnetic-particle inspection of cross sections cut from the specimen.

#### Restrained Fillet-Weld Lap Joint

A restrained fillet-weld lap-joint was developed by Swinden and Reeve<sup>(H275)</sup> as a means of evaluating the effect of changes in welding variables on the cracking tendency of welds in low-alloy structural steels. It is widely used in Great Britain to study both weld-metal cracking and underbead cracking in the parent metal and is commonly referred to as the Reeve test.

The Reeve test consists of two superimposed test plates bolted to an anchor plate, as shown in Figure 48. Restraint, in addition to that supplied by bolting, is provided by anchor fillet welds in three of the four fillet joints. The test fillet is deposited after first allowing the plates to cool to room temperature.

The effect of a change in a variable on cracking tendency is determined by the amount of metallographic preparation of a section of the weld which is necessary to make the crack visible. For instance, during a test, a crack may be visible immediately after welding is completed. This would be a very severe crack and would be classified as a Class I crack. The next test made with a change in a welding variable may reveal a crack only after a section of the weld is cut and rough polished. Thus, it would be said that the change made in the variable causes a decrease in cracking susceptibility because the resulting crack is a Class II crack which is visible only after cutting and rough polishing. This test is adaptable to any thickness of plate and is sensitive to changes in any of the welding variables.

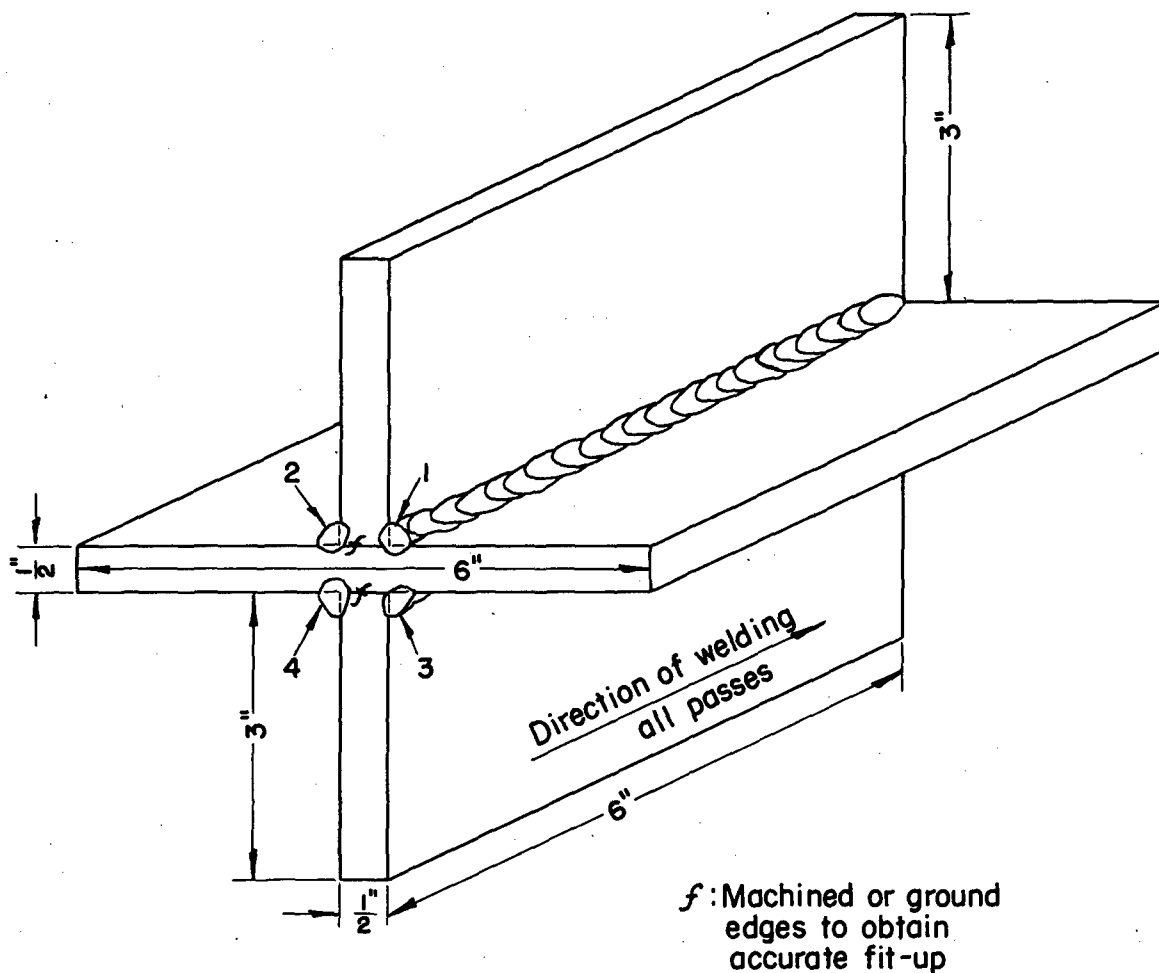


FIGURE 47. SPECIMEN FOR FILLET-WELDED DOUBLE-TEE-JOINT TEST

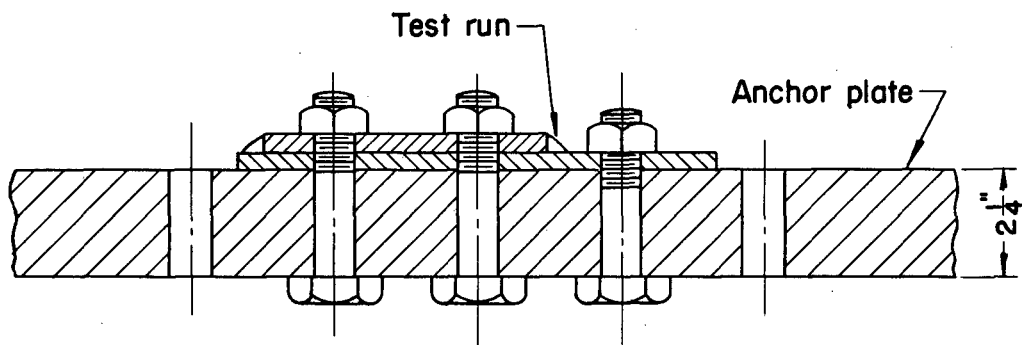
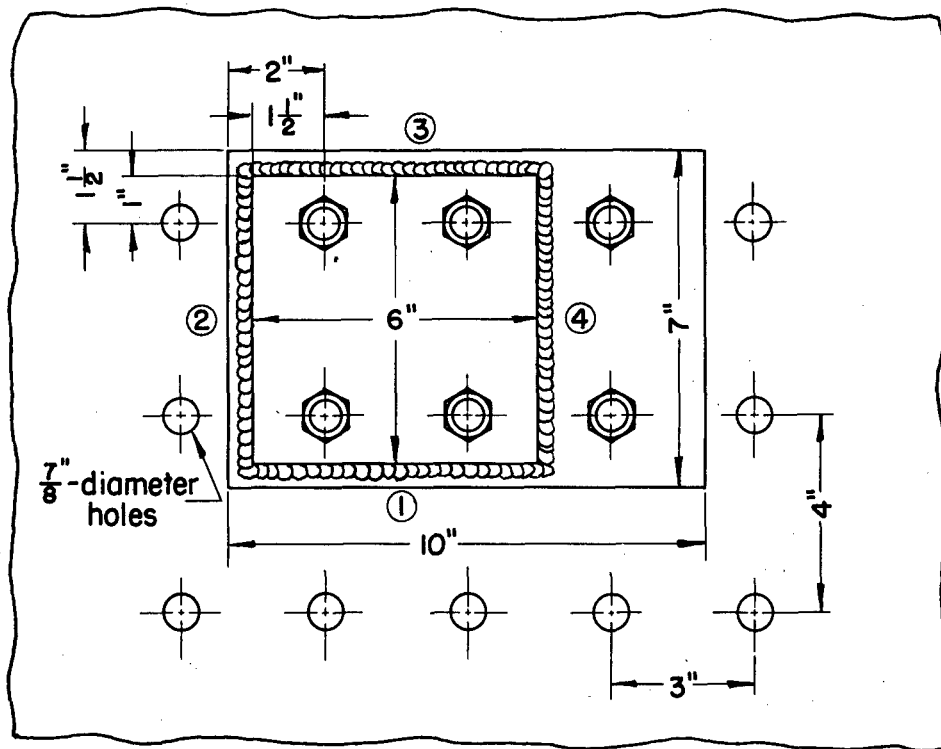


FIGURE 48. RESTRAINED FILLET-WELD LAP-JOINT TEST (H275)

It is conceded, however, that this test falls short in that no provision is made to determine the degree of restraint necessary to cause a crack. The test is designed so that all test welds are made under full restraint. This brings out another advantage of the test, however. That is, only one test run is needed under any one set of conditions to determine the severity of cracking.

In evaluating susceptibility to cracking in tests when specimens vary only slightly in the severity of cracking, the dependency on the human element for evaluating the severity of cracking would be a shortcoming. It seems that, in such cases, the more recent Mond and Lehigh tests as discussed below would be more reliable, due to their greater sensitivities.

### Restrained Butt-Joint

Investigators in the literature have used various types of butt-joint weld tests to study both weld-metal cracking and base-metal cracking. Muller<sup>(H184)</sup> developed a butt-joint test to study hot cracking in welds made on chromium-molybdenum steel sheet. In this test, steel sheets are rigidly clamped with bolts so that the abutting edges are separated by a distance equal to the thickness of the sheet. Muller used this test primarily to study hot cracking which was encountered in the base metal adjacent to welds deposited with the oxyacetylene welding process. However, the same test is often used to study weld-metal cracking.

At Battelle<sup>(H58)</sup>, a restraining weld fixture was designed especially for studies of cracking in metal-arc welds deposited on aircraft steel plate 3/16 inch to 3/4 inch thick. This fixture is shown in Figure 49. The specimen, a single-vee butt joint with a backup, is rigidly bolted to the fixture to minimize distortion. The flat bars (F), welded to the base plates, prevent movement of the plates transverse to the direction of welding. The specimens were also preheated by means of the strip heaters shown in Figure 49.

A restrained butt-weld cracking test was developed by Swinden and Reeve<sup>(H275)</sup> during an investigation of cracking in structural steel weldments. It is somewhat similar to their restrained fillet-weld-cracking test developed by the same investigators, except that the test weld is a highly restrained butt weld. The assembly is shown in Figure 50. The entire assembly is cooled to room temperature before the final test weld is deposited.

Weld-metal cracking is more prevalent with this test than it is with the fillet-weld test. This test was, therefore, recommended for studies of weld-metal cracking.

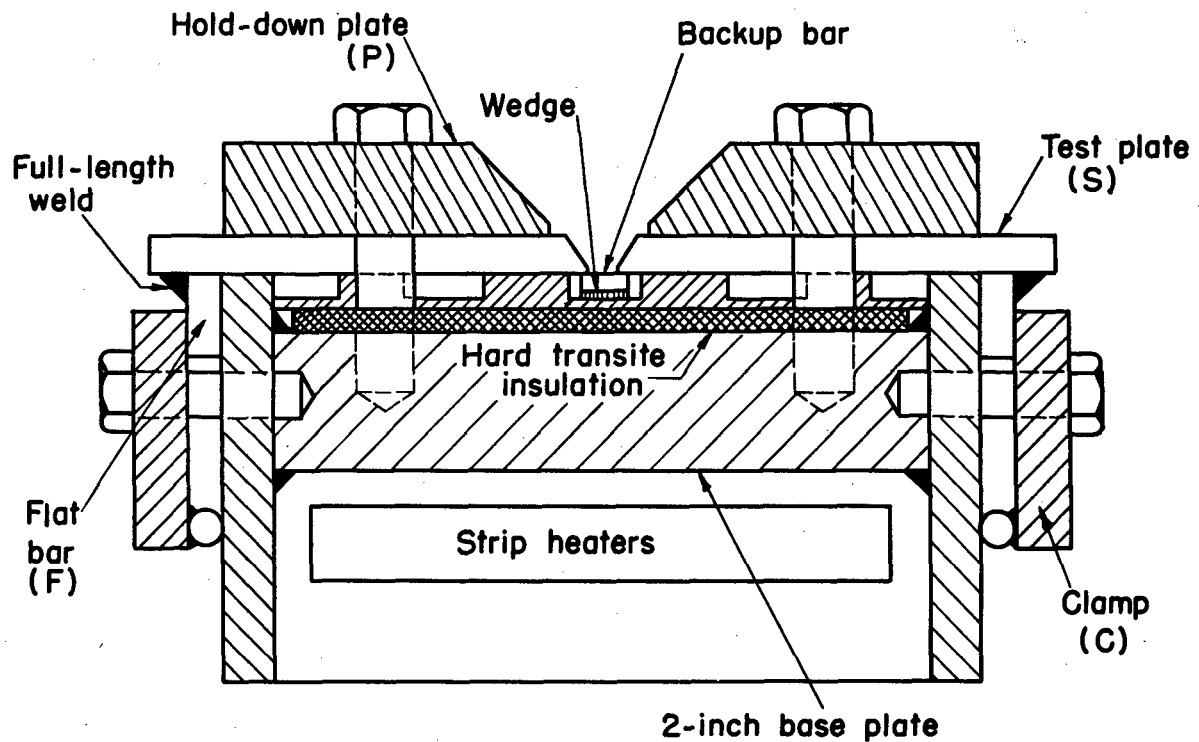


FIGURE 49. SCHEMATIC SKETCH OF RESTRAINING WELD FIXTURE

#### "X"-Weld Cracking Test

The "X"-weld cracking test is used to study cracking in weld metal deposited with nickel-chromium-iron alloys<sup>(117)</sup>. The test assembly is shown in Figure 51. Two 3-inch-long pieces cut from 1-inch square bar stock are clamped with their edges together to form a double-vee groove. Passes are laid in, two at a time, on alternate sides of the double vee, allowing the specimen to cool to about 100°F, or lower, between each pair of passes. With 5/32-inch electrodes, a total of 12 passes is normally employed using the bead sequence shown in Figure 51A.

The degree of crack susceptibility is determined by observing when the cracks first appear during welding. The most sensitive welds crack after the third pass. Specimens which show no cracks during welding are then sectioned and polished for closer examination. The appearance of a badly cracked weld is shown in Figure 51B.

This specimen's main advantages are its simplicity, reproducibility, and low cost.

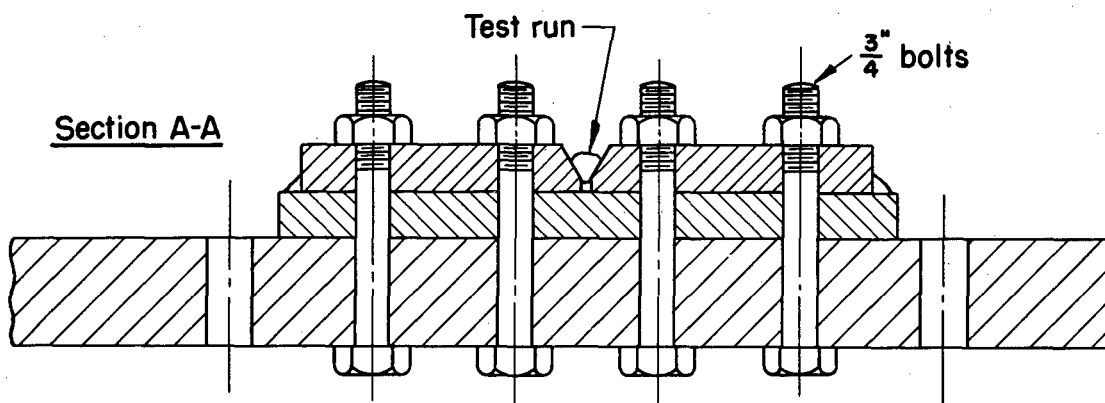
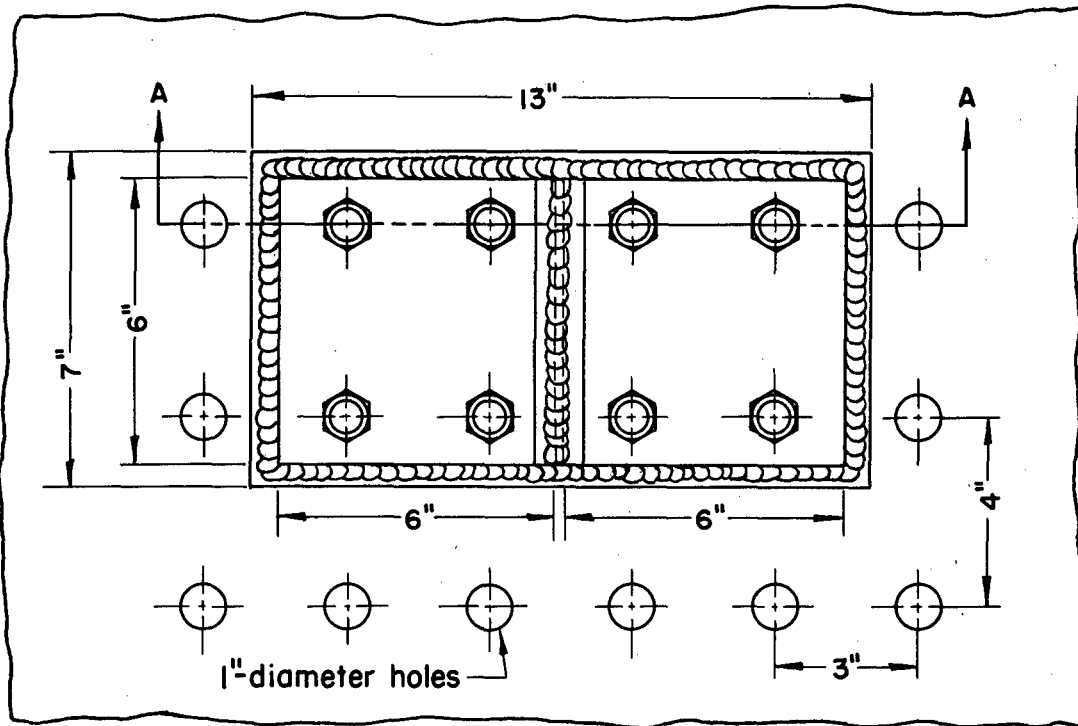
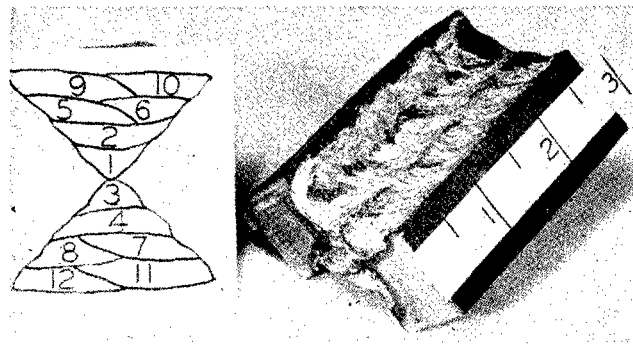


FIGURE 50. RESTRAINED BUTT-WELD CRACKING TEST<sup>(H275)</sup>



A. Welded Specimen and Bead Sequence



B. Appearance of Cracks in Weld Metal

FIGURE 51. "X"-WELD CRACKING-TEST SPECIMEN SHOWING BEAD SEQUENCE AND APPEARANCE OF CRACKS<sup>(117)</sup>

#### Circular-Patch Test

As discussed previously in this report, one of the most difficultly welded joints is encountered when inserting a patch in a plate or boiler. The circular-patch test shown in Figure 52 was developed to simulate this severe type of joint. It consists essentially of a root pass deposited in a circular double-vee butt joint.

In conducting this test, the center patch is tacked in two places in a central position and then one run of weld metal is deposited in the groove. The weld is deposited slightly short of the tacks and the craters are left open. The extent of cracking is measured and expressed as a percentage of the total circumferential length of weld metal.



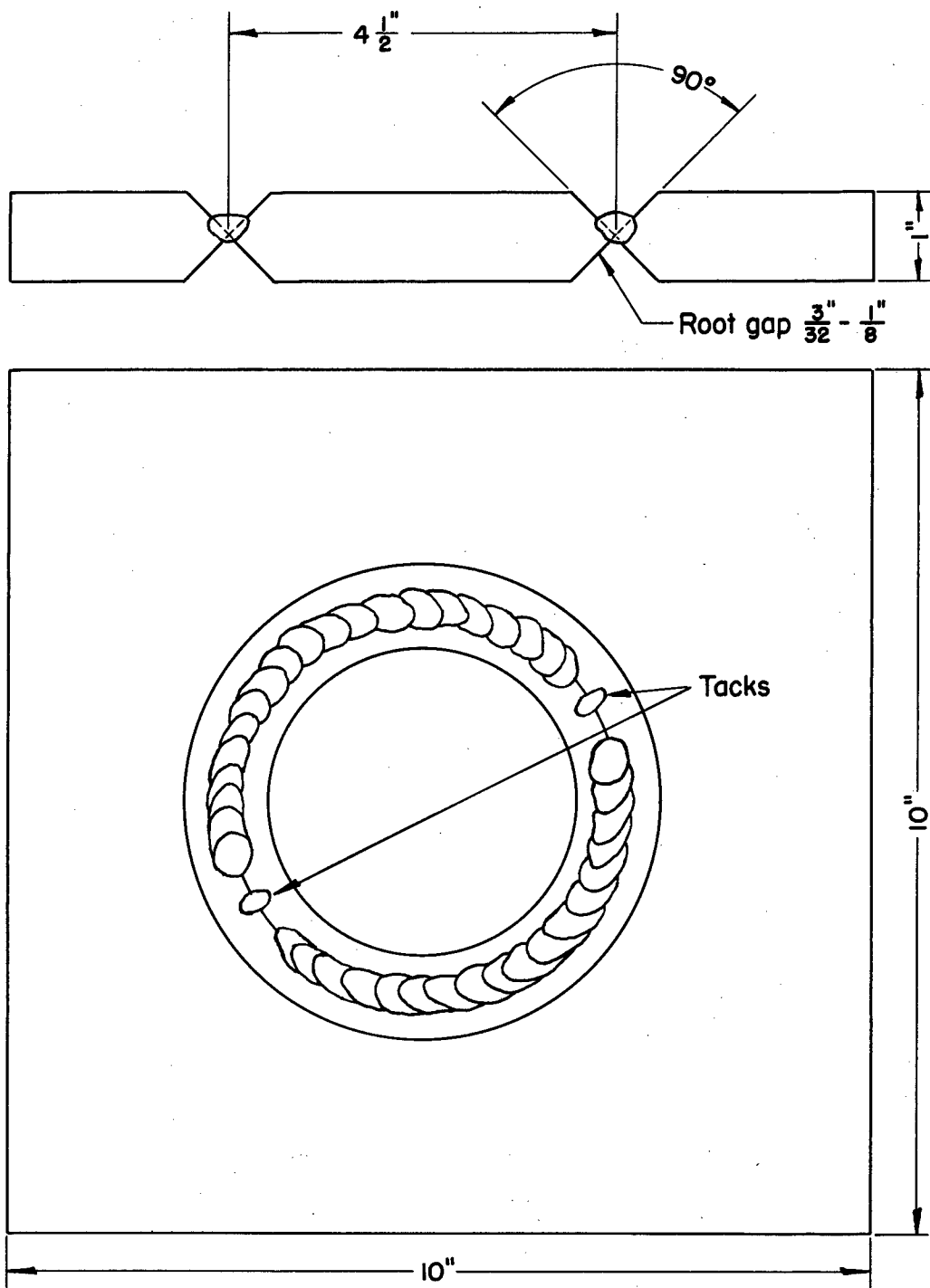


FIGURE 52. CIRCULAR PATCH-TEST SPECIMEN

### Circular-Groove Test

The circular-groove test specimen was first introduced in cracking studies of welds made in high-temperature alloys<sup>(151)</sup>. This specimen, or modifications of it, has since been used by various investigators to evaluate the cracking susceptibility of all types of weld metal.

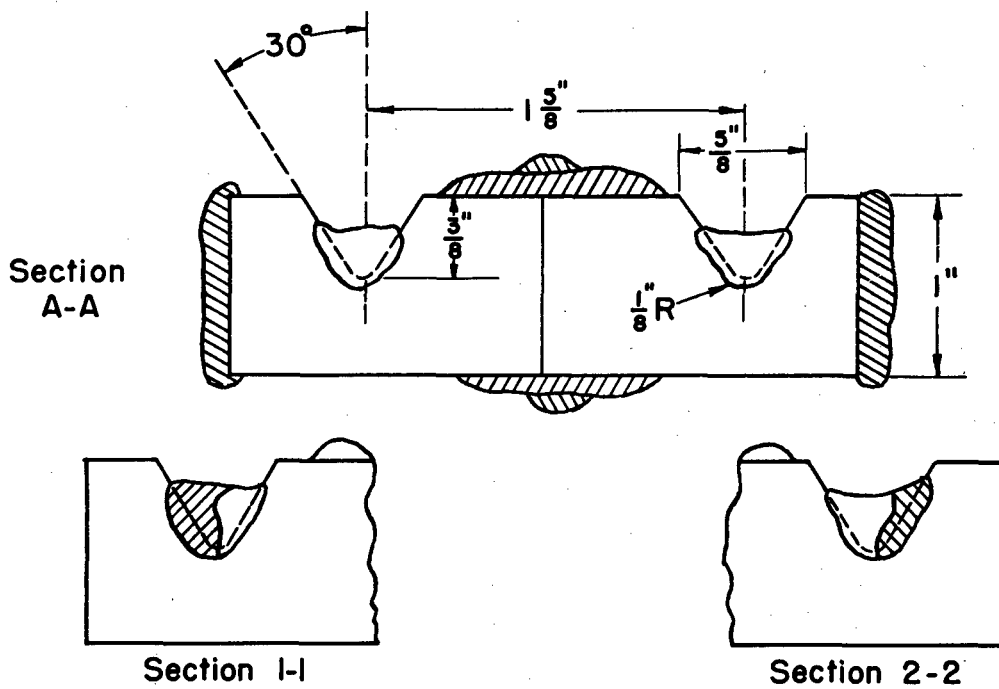
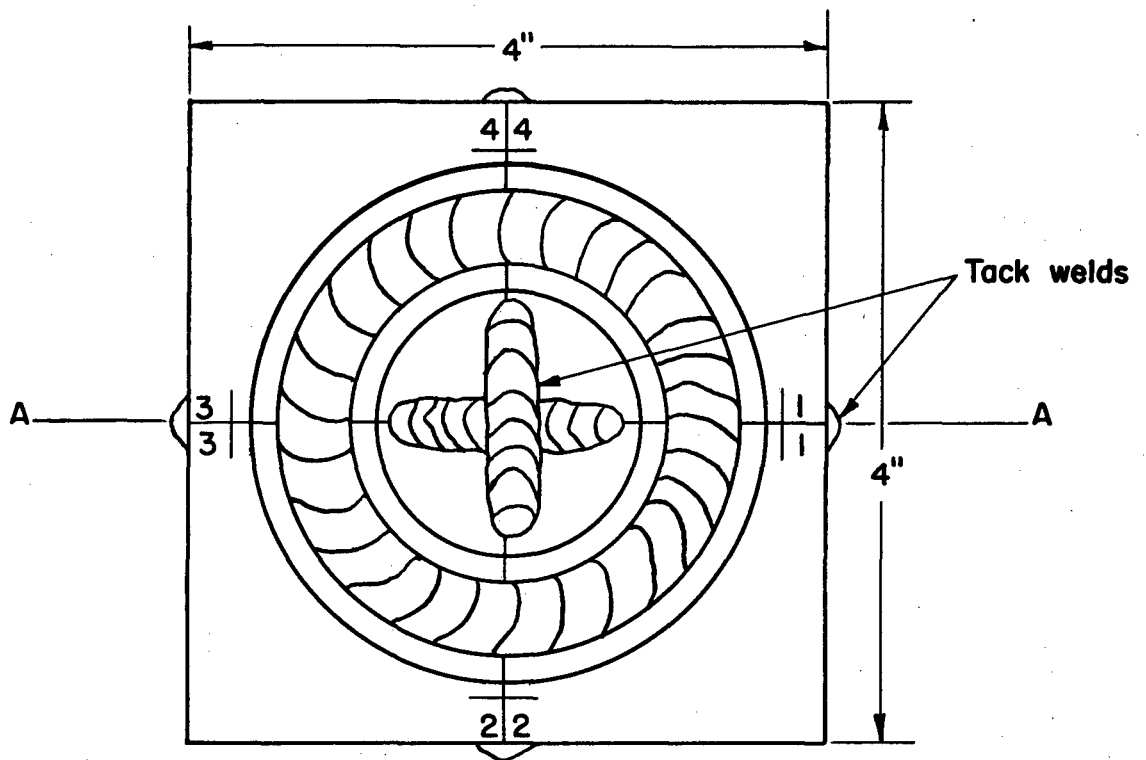
The specimen consists of a simple circular groove with tapering scarves and a rounded bottom cut into the surface of a square piece of plate. A photograph of a cracked weld deposit in a circular-groove specimen is shown in Figure 31 (earlier in this report). This type of groove eliminates notch effects due to butting plates and backing strips. This specimen shows the cracking due to restraint alone.

A modified version of the circular-groove specimen is shown in Figure 53. This specimen is cut into four segments to provide an interface notch effect at the junctions. The segments were finish machined on the abutting edges and then tacked together before depositing the weld. The test was used to simulate the interbucket extension cracking encountered in welding buckets to gas-turbine discs. The extent of cracking is determined by breaking apart the segments and noting the cracked (oxidized) area on the weld fracture.

### Mond Cracking Test

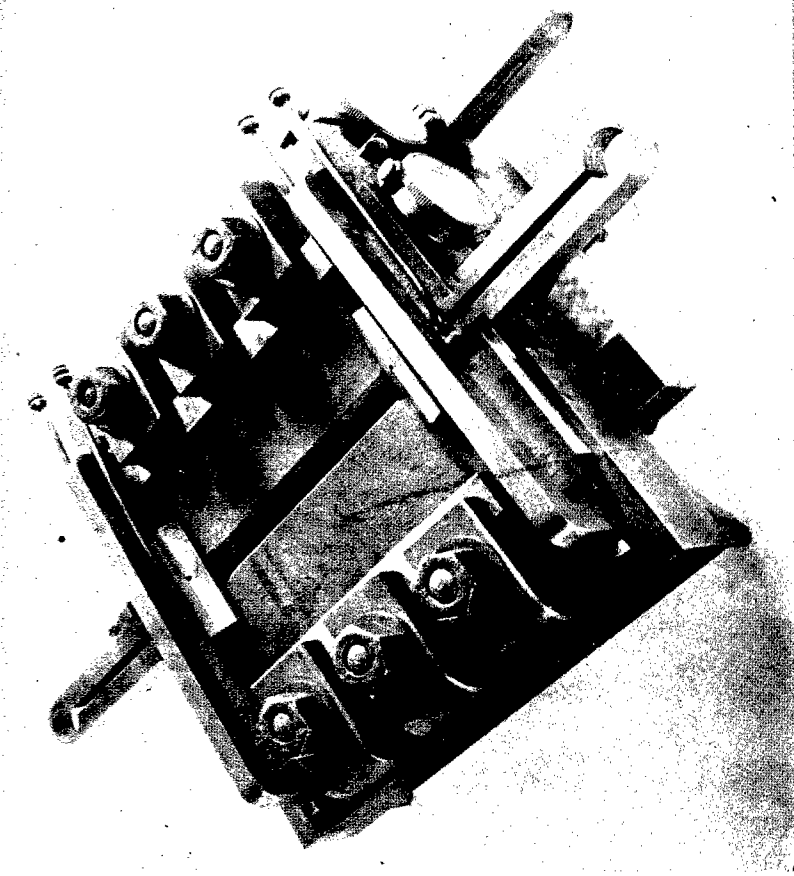
The Mond cracking test was developed in Great Britain by the Mond Nickel Company, Limited, in order to make quantitative studies of stress in welds and its relation to weld cracking<sup>(H165)</sup>. The test is especially designed to study under-bead cracking in welds, but is also applicable to weld-metal cracking studies. Photographs and detailed drawings of the test setup are shown in Figures 54 and 55 respectively.

Details of the testing apparatus may be found in the author's paper<sup>(H165)</sup>. However, the test essentially consists of a butt-weld specimen (Figure 55e) which is restrained in a calibrated jig during welding. The jig restrains the movement of the test plate in two directions. Compression cylinders with dial gages, Figure 54b, are attached to each end of the butt-weld joint. The compression of these cylinders provides a direct measure of the main force across the weld. A third compression cylinder is used to measure the deflection of the jig due to the bending moment produced during welding. The compression cylinders were constructed of hardened high-carbon nickel-chromium steel. They were calibrated in a testing machine. The authors stated that the calibration was only correct prior to welding. Readings were therefore approximate.

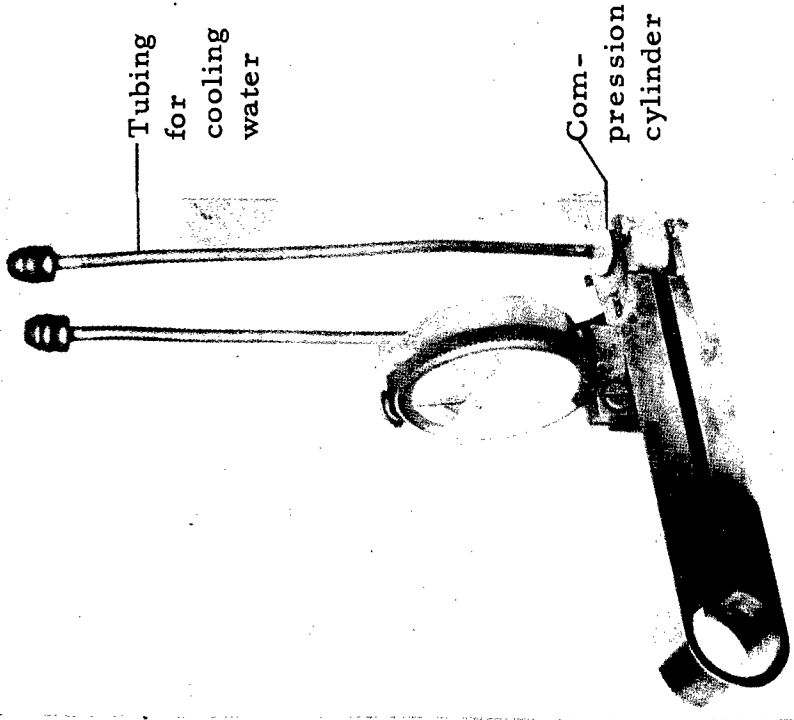


Shaded areas show extent of weld cracks in plane of interblock junction

FIGURE 53. CIRCULAR-GROOVE TEST SPECIMEN (I51)



A. Complete Jig Assembly



B. Compression Cylinder  
Attachment and Dial Gage

FIGURE 54. PHOTOGRAPHS OF MOND WELD-CRACKING JIG AND STRESS MEASURING ASSEMBLY (H165)

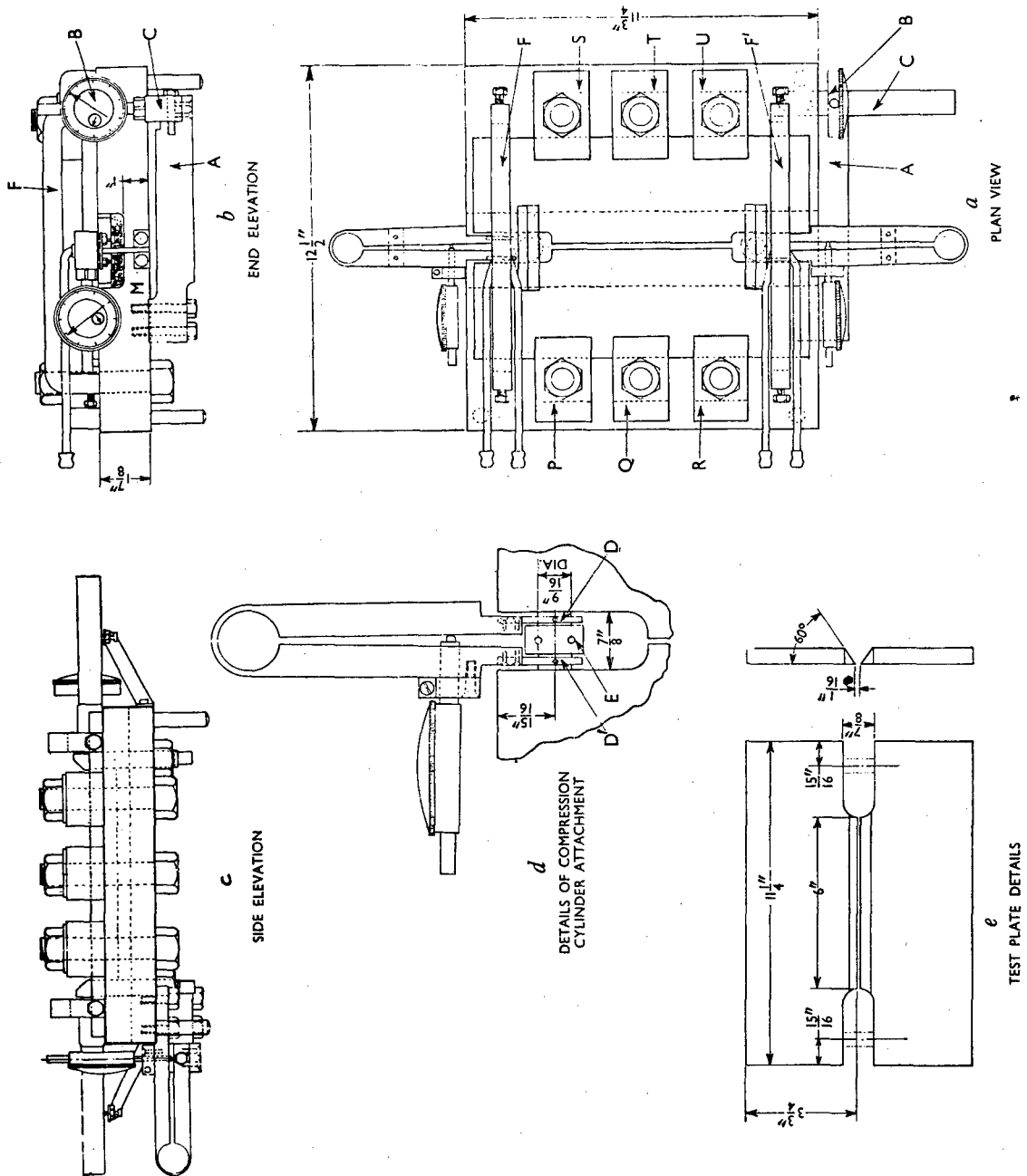


FIGURE 55. DETAILS OF MOND WELD-CRACKING JIG AND COMPONENTS(H165)

The initiation and propagation of a weld crack is detected by the drop in indicated stress which the crack produces. Stress measurements are made indirectly by recording the deflection of the calibrated components with respect to time. Stress-time curves are obtained from these results. Temperature-time curves are also obtained by means of a thermocouple and a recorder.

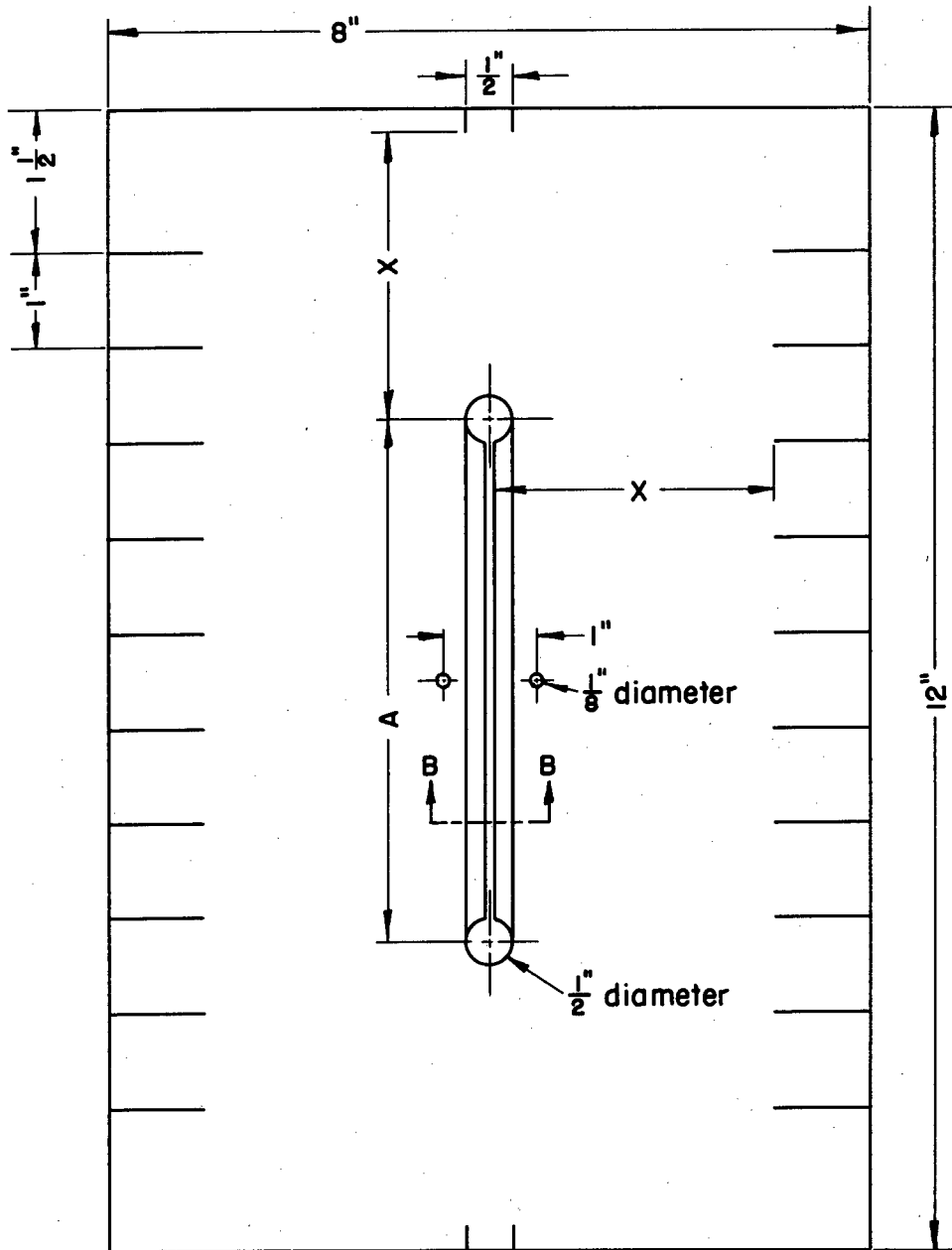
This test is very effective in studies of under-bead cracking in welds. It may be equally effective in determining the stress levels and temperatures at which weld-metal cracking occurs. So far, the Mond test has not been used in studies of weld-metal cracking. In such studies, the welding electrodes would be the variable instead of the base-plate material.

### Lehigh Restraint Test

The Lehigh restraint test was developed for evaluating quantitatively the degree of restraint at which cracking occurred during cooling<sup>(H271)</sup>. The specimen is shown in Figure 56. Slots are cut into the sides and ends. By changing the length of the slots, the degree of plate restraint is varied from one specimen to another. This restraint is expressed numerically by the width of the specimen between the bottom of the slots, measured in inches or twice the dimension "X" in Figure 56. The threshold degree of restraint for cracking is expressed as that width (measured to the nearest half inch) which is just sufficient to cause cracking. The groove geometry is held constant except for necessary variations in its length. For 1-inch and 1- $\frac{1}{2}$ -inch plates, the groove length is 5 inches, but for  $\frac{1}{2}$ -inch and  $\frac{3}{4}$ -inch plates, the groove length is 3 inches, in order to provide sufficient restraint to cause cracking in thin steel plate.

Cracks are detected by microexamination of weld cross section and also by measuring the transverse contraction at the mid-length of the weld. With this latter procedure, it is only necessary to section and examine those specimens exhibiting slightly more transverse contraction than that known to be associated with cracking. In general, only weld-metal cracks are observed with this test. However, in steels particularly susceptible to base-metal cracking, such cracks were observed upon cross sectioning.

In addition to restraint, the Lehigh specimen is useful in studying the effects of weld-metal composition, heat input, preheat, and differences in classes of electrodes, on the cracking tendency. The test has the disadvantage that the specimens are expensive to machine and several of them must be used to determine the cracking threshold under a single set of conditions.



A  $\begin{cases} 3\frac{1}{2}'' \text{ for plate } < 1'' \\ 5\frac{1}{2}'' \text{ for plate } \leq 1'' \end{cases}$

Section B-B

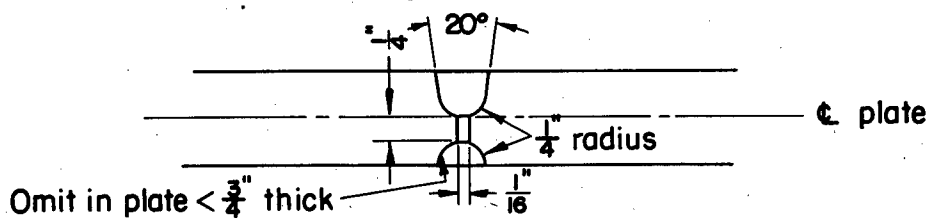


FIGURE 56. LEHIGH FIN-TYPE CRACKING-TEST SPECIMEN (H271)

## Naval Research Laboratory Specimen

This specimen was developed by the Naval Research Laboratory as a less expensive substitute for the Lehigh fin-type specimen. It is an attempt to overcome the necessity of preparing a series of specimens, since only one specimen is necessary to determine the threshold of cracking for a given material and welding procedure. The NLR specimen imposes varying degrees of restraint along the length of the weld. The specimen is shown in Figure 57.

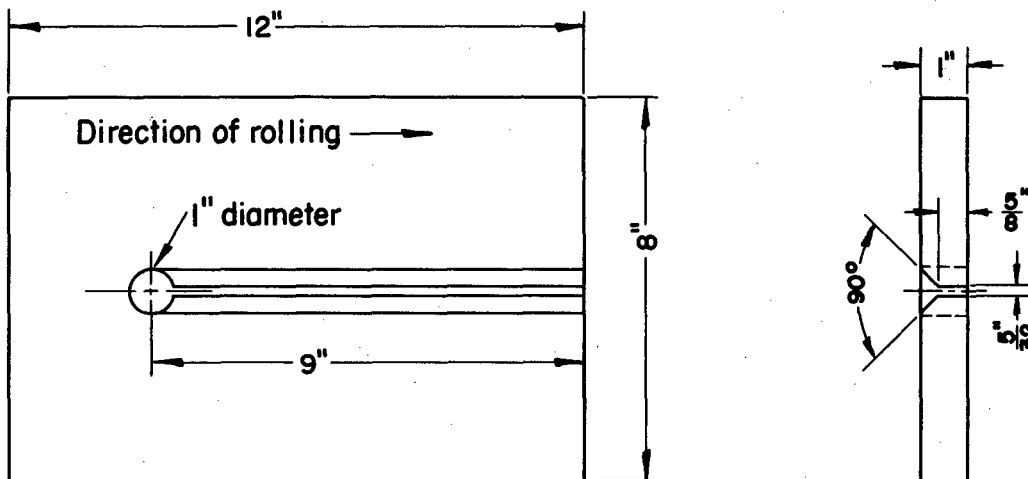


FIGURE 57. NAVAL RESEARCH LABORATORY  
CRACK-SENSITIVITY SPECIMEN

The weld is deposited starting at the edge of the plate and progressing inward toward the drilled hole. As the weld progresses, the amount of restraint increases. Weld-metal cracking starts at the drilled hole and progresses outward. The length of cracking is taken as the index of crack sensitivity under these conditions of restraint. The crack length also indicates the relative stress necessary to cause cracking.

Initial studies with this specimen at the Naval Research Laboratory showed that the results were not reproducible. Apparently, residual stresses induced into the specimen during machining caused difficulties in insuring a uniform groove geometry.



## Hot-Cracking Machine

Murex Welding Processes, Ltd., developed a cracking test in which an external bending force is applied to a fillet-weld lap joint during welding<sup>(H178)</sup>. The test specimen consists of two plates of 1/2-inch-thick mild steel, 2 inches wide and 3-1/2 inches long, which are placed in a bending machine before welding. The lap joint is one in which the lapping members have been moved back so that the surfaces are no longer in contact as shown in Figure 58. The weld is made on the top fillet only. A gap of 1/32 inch is allowed between the edges.

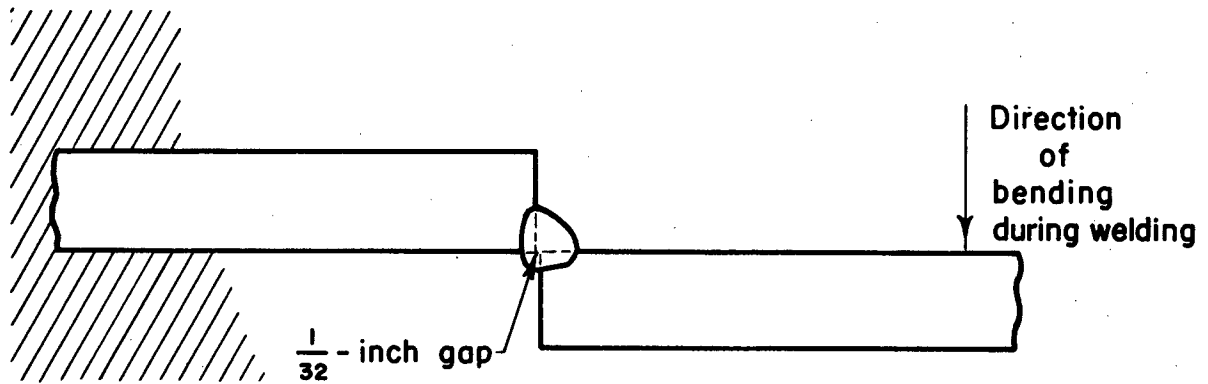
During the actual welding operation, the lower plate is steadily bent in a clockwise direction, with its top corner as the center of the radius. The upper plate is held fast in a vise. Bending is accomplished by means of a worm and quadrant driven at a predetermined speed (not mentioned) with a motor-driven variable-speed gear. The welder starts welding at one end of the lap joint and, after 10 seconds delay, he engages the clutch to put the bending gear in operation. After welding for about 30 seconds, standardized for each size electrode, the welding operation and bending movement is stopped.

As a result of the bending action, longitudinal cracks appear in the weld-metal deposit. The degree of cracking susceptibility of a particular weld is determined by measuring the length of these cracks.

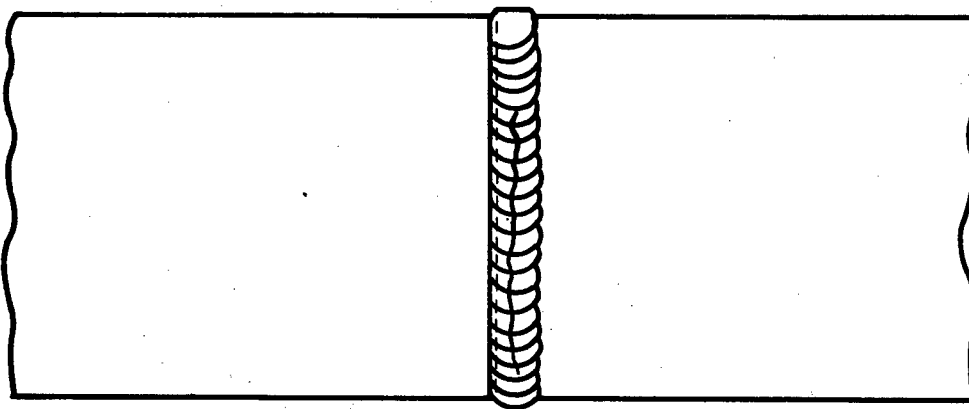
## Related Tests

Several investigators have studied the mechanical properties of weld metal at high temperatures in order to correlate strength and ductility with cracking susceptibility. The results of such tests by various investigators were presented earlier in this report when discussing the metallurgical properties of weld metal.

At Battelle<sup>(I51)</sup>, hot-ductility tests were made using unwelded specimens and transverse weld-joint specimens of high-temperature alloys. The dimensions of these specimens are shown in Figure 59. The specimens were tested in the assembly shown in Figure 60. The specimens were enclosed within a Pyrex-glass tube filled with dehydrated Sil-O-Cel powder. They were heated by passing an electric current through them. The heating time was one to two minutes to about 2400°F. Purified argon was used as a protective atmosphere. Temperatures were measured by means of a platinum - platinum-rhodium thermocouple spot welded to each specimen. It was found necessary to wrap the thermocouple lead wires around the specimen several times to insure heating of the thermocouple bead to the same temperature as the specimen. The specimens were



A- Loading arrangement (schematic)



B- Appearance of cracked weld after welding

FIGURE 58. LAP-JOINT WELD SPECIMEN FOR  
HOT-CRACKING MACHINE<sup>(H178)</sup>

(Drawn from author's description)

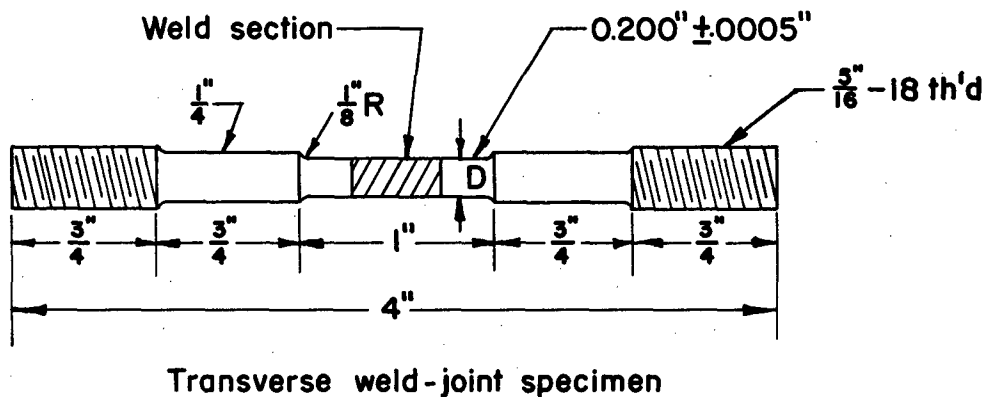
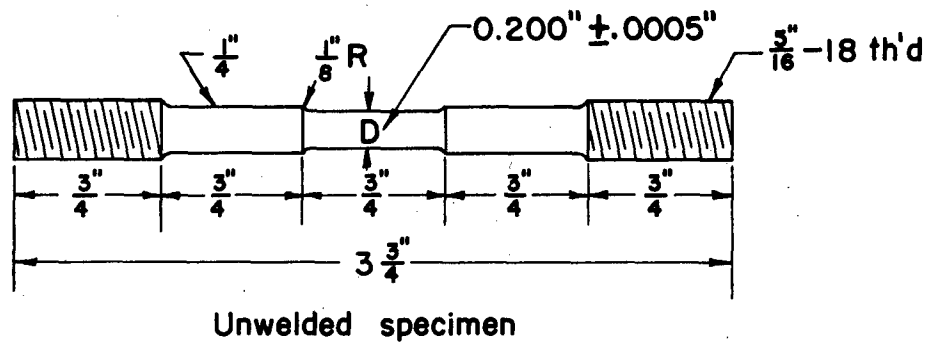


FIGURE 59. HOT-SHORTNESS TEST SPECIMENS<sup>(151)</sup>

pulled apart after 5 minutes at temperature, using a testing apparatus similar to one used later by Rollason, as shown in Figure 61<sup>(H211)</sup>. This investigator also heated his specimens by passing current through them. He employed an optical pyrometer instead of using thermocouples for measuring temperatures.

Ball and Winterton<sup>(H25)</sup> made hot tensile tests of similar specimens cut transverse and longitudinal to the direction of welding. They heated their specimens in a platinum-wound furnace. A possible disadvantage to this method is the longer heating time required—10 to 15 minutes to reach temperatures of 1850°F to 2000°F.

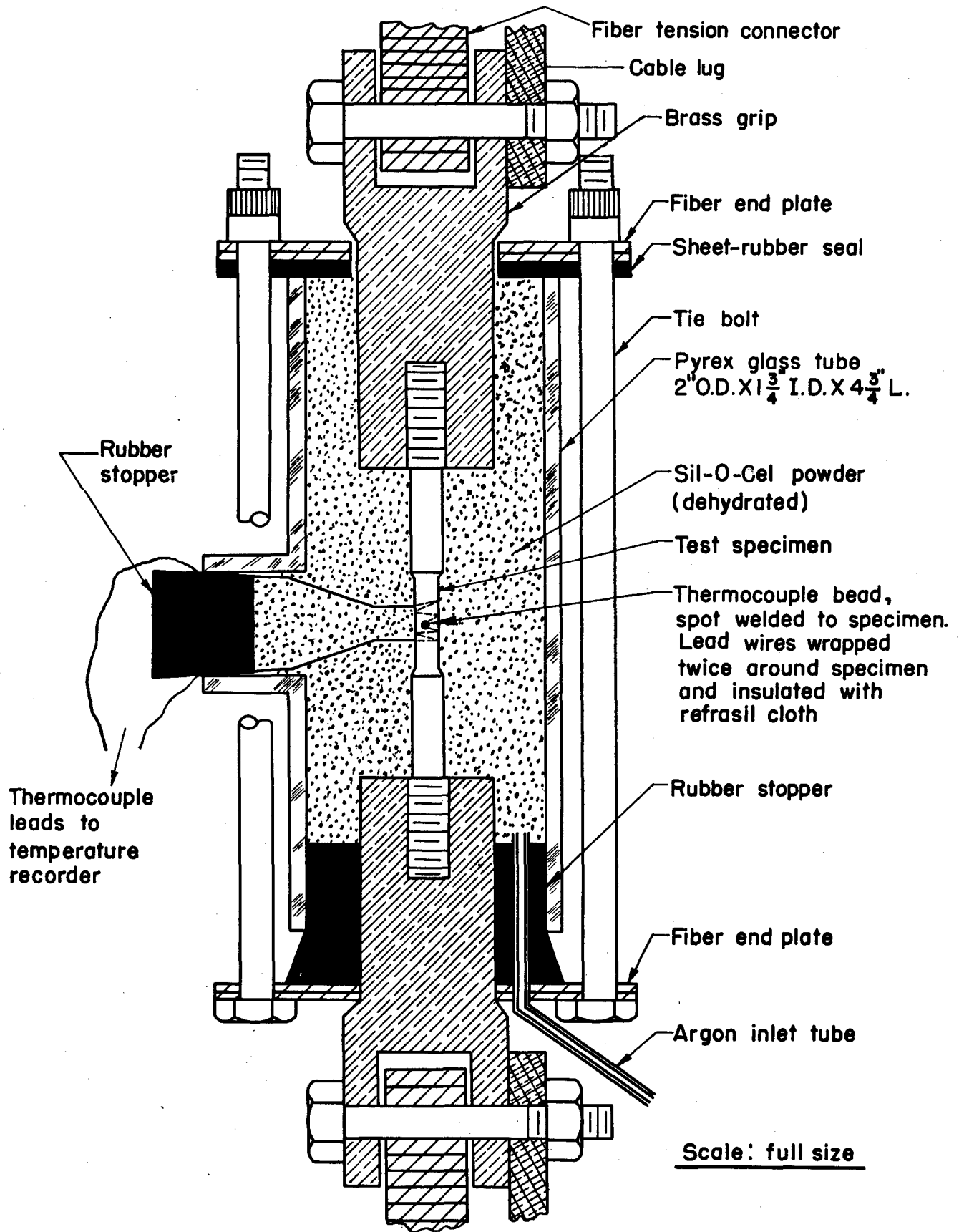


FIGURE 60. HOT-SHORTNESS TEST ASSEMBLY (151)

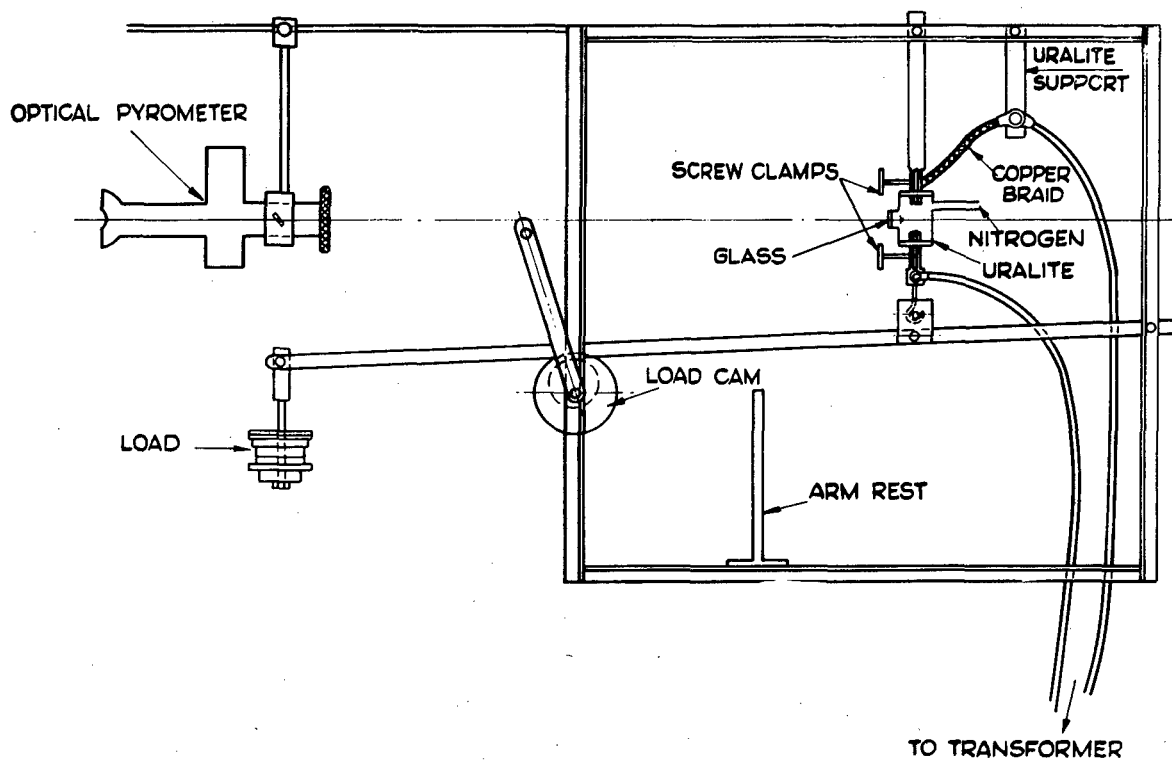


FIGURE 61. HOT TENSILE-TEST APPARATUS(H211)

## SUMMARY

This literature survey has supplied background information on cracking in weld metal. The pertinent points are summarized in the following statements:

1. The major type of cracking associated with weld metal is hot cracking, which occurs at elevated temperatures.
2. Little reliable information is available on the temperature at which hot cracking occurs. The methods used, so far, for determining it have not proved to be very satisfactory.
3. Most weld-metal cracking is interdendritic or intergranular, regardless of the temperature at which it is believed to occur.
4. Hot cracking is generally believed to be related to hot tearing, although hot shortness may also be a factor.
5. In ferritic weld deposits, elements such as sulfur, phosphorus, oxygen, nickel (in combination with sulfur), and carbon increase susceptibility to hot cracking. Manganese reduces the tendency to hot cracking.
6. In austenitic weld deposits, elements which promote the formation of delta ferrite tend to reduce the cracking tendency. Chromium, molybdenum, columbium, and silicon induce delta-ferrite formation. Nickel, manganese, nitrogen, and carbon suppress its formation. Silicon and columbium, however, have an unfavorable effect on cracking susceptibility of strongly austenitic weld metal.
7. In addition to composition, metallurgical factors such as segregation and dendritic grain size affect crack susceptibility.
8. Preheat and postheat appear to have little effect in controlling hot cracking in weld metal and, in some cases, may tend to increase the cracking. However, cold cracking can be prevented by preheating and postheating.
9. High-temperature tensile tests of weld-metal specimens (reheated from room temperature) show a slight relation to cracking tendency, but the results are not conclusive. Tensile tests with weld-metal specimens made during cooling from the melting temperature were not mentioned in the literature.

10. The circular-patch and circular-groove tests are good all-around tests for making quantitative studies of the hot cracking tendency of weld metal.

Since most weld-metal cracking is interdendritic or intergranular, it would be well to review factors which could cause brittle grain-boundary failures in weld metal during cooling from the welding temperature. This predisposition to fail locally may be due to two factors:

1. Presence of molten eutectic, or liquid impurities, at grain boundaries.
2. Grain-boundary weakness with or without a grain-boundary phase or impurity.

Since all known types of weld metal will crack if the conditions are correct, it appears that the stress conditions are of critical importance in governing cracking. This is borne out by experience. Furthermore, it is a matter of record that some weld metals are more prone to crack than others. Consequently, weld-metal cracking is a function of both the welding condition and the weld metal.

It would greatly assist and simplify the solution of the problem of weld-metal cracking if more specific knowledge were available to correlate stress conditions and weld-metal characteristics with the cracking tendency. Perhaps the simplest item of information is the temperature at which the cracks start — yet data on this point upon which full reliance can be placed are very meager if, indeed, any are available at all. The establishment of such information, for at least representative cases of stress and weld metal, would go a long way towards defining the effective hazards.

With the highest purity metal available at present, it appears that a slowly applied stress is capable of pulling a metal apart at grain boundaries when the temperature is below the freezing point, but above some relatively poorly defined "annealing" or "equi-cohesive" temperature. Naturally, that raises the question of why we do not always find cracks in weld metal that freezes and cools under restraint. This is a very inadequately explored field, but it could be that certain "impurity" elements are needed to strengthen the grain boundaries under welding conditions.

Actual weld metal contains alloys and impurities and there is information on the effects of such elements on hot tearing and hot shortness. However, to correlate the presence of such an element with cracking, the temperature of crack formation should be known. Judging from this survey it is expected that hot shortness is of relatively minor importance. Hot tearing seems to be the more likely cracking process. This should be checked by future research. For example, if FeO were known to be the principal impurity of a weld metal and cracking were known to occur at some temperature from 2200°F up to the freezing temperature (solidus), hot tearing rather than hot shortness would be indicated as the cracking process.

This comes from the finding that it is at the 1800 F temperature range that oxygen as FeO makes steel hot short. In principle, the same could be established for S, or FeS, though hot shortness from S comes at a higher temperature and the experimental difficulties of making the distinction would probably be somewhat greater. Thus, it seems likely that a weld-metal crack may start at some temperature definitely below the solidus at which the stress generated by shrinkage is sufficient to overcome the strength or cohesion of the grain boundaries.

A further reason for needing to know the cracking temperature is the postulated effect of a liquid of some kind at the grain or dendrite boundaries. Grain-boundary constituents should be identified so that their melting points can be established. If the melting points of these constituents correspond with the cracking temperature, then it may be concluded that liquid films are a cause of the cracking. It may be that special alloying elements could be added to raise the melting point of the grain boundary material.

The experimental work on this investigation will be concentrated on studies of various factors which were not clearly established in the literature. The initial studies will be made of dendritic grain boundaries of crack-sensitive weld metal to determine what constituents, if any, are causing the weakness which leads to failure by cracking. Microstructures of the grain boundaries will be studied with the optical microscope and the electron microscope. The studies will also involve electron-diffraction techniques and microspectrographic analysis as an aid in identifying the constituents in the grain boundaries. In the electron-diffraction studies, the grain boundaries will have to be brought up in relief by special etching techniques so that the electron beam can be transmitted through them. This type of study is now being made by other investigators to identify grain-boundary constituents in austenitic weld metal.

An indirect approach to the problem of the effect of impurities is to start with a pure material and then add various elements to study their effect. Plans are now being made to prepare high-purity SAE 4340 steel base plates and high-purity high-strength filler-wire compositions. If welds made with the pure materials are crack free, then studies will be made of the effects of various impurities by adding them to the weld deposits.

Tests will be made to determine temperatures at which weld-metal cracking occurs in ferritic welds. High-temperature strength and ductility tests will be made both on cooling from the melting temperature and on heating from room temperature. Techniques are also being considered for making direct measurements of cracking temperatures using precise methods for detecting crack initiation.



APPENDIX

## APPENDIX

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