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WELDING OF CHROME-NICKEL AUSTENITE STEELS

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6. Porosity in Welds of Chromium-Nickel Austenitic Steels

The principal cause of porosity in welds in carbon and low-alloy steels is hydrogen. Nitrogen may also cause pores during the welding of ordinary steels. This is shown by the well-known fact that pores appear during submerged-arc welding when air gets into the fusion zone (welding with a gap between the corner welds of tee joints; unilateral welding of overhanging butt welds with insufficient penetration, welding when the flux burden level is insufficient, and so on).

Experiments carried out at the Institute of Electric Welding have probed that when austenitic steels are welded, hydrogen is mainly responsible for the formation of pores; nitrogen and oxygen hardly have any such effect. We have already pointed out that when these steels are arc welded there is practically no oxidation of the carbon. In view of this, the problem of the effect of carbon monoxide on pore formation in austenitic welds loses all practical importance*. Consequently, we are concerned basically with one gas alone -- hydrogen. Experiments were carried out in which carbon and stainless steels were welded under the same conditions with a supply of nitrogen to the combustion zone. When equal amounts of nitrogen were supplied

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* In welding in an atmosphere of CO_2 pores may appear due to the generation of carbon monoxide from the crystallizing weld pool.

the carbon weld was riddled with pores and blowholes, while the austenitic weld was completely solid.

Fig. 82 shows the surface of the two welds. The almost incomplete immunity of chrome-nickel austenitic steel welds to porosity caused by nitrogen can be explained by the excellent solubility of nitrogen in these steels /143/. It is just this fact which has made it possible to carry out the so-called nitrogen-arc welding of stainless steels (work of A. Ya. Brodskiy and others).

Fig. 82. Effect of nitrogen on porosity of steel welds:

a) St. 3; b) 1Kh18N9T.

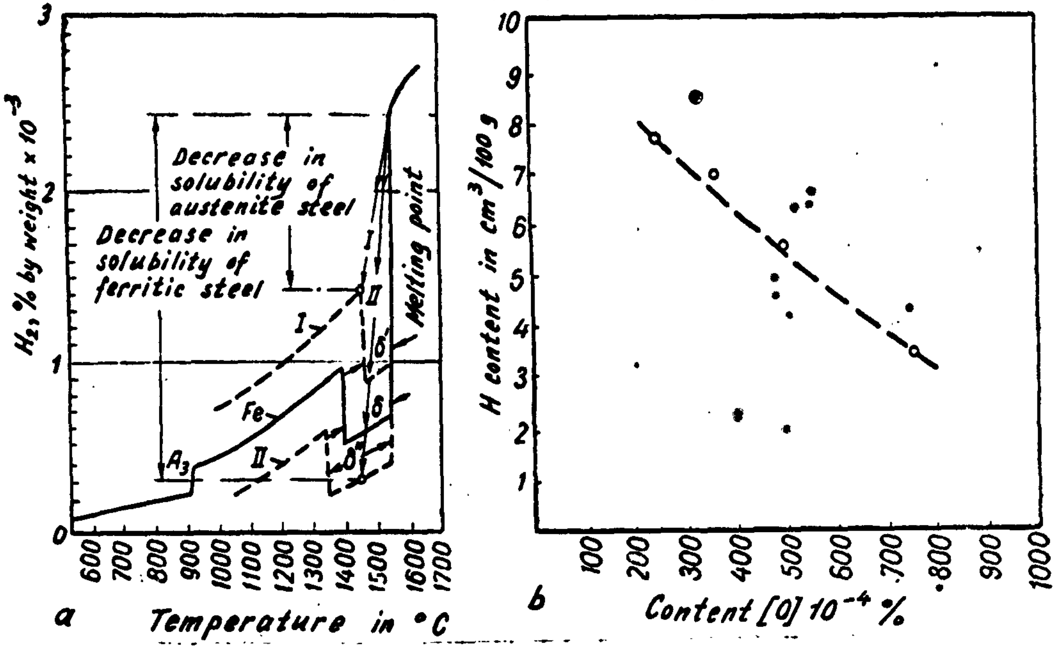

Fig. 83. (a) variation in the solubility of hydrogen during solidification of austenitic (I) and ferritic (II) steels; (b) relation between content of oxygen and hydrogen in a 25-20 steel weld.

The formation of pores is associated with a sudden, jerky decrease in the solubility of gases in the metal at the instant of crystallization of the weld pool. When austenitic steel solidifies, the fall in the solubility of hydrogen is considerably less than in the case of conventional low-carbon and ferritic steels (Fig. 83).

Hence austenitic steel welds are less damaged by porosity than ordinary steel ones.

The main sources of hydrogen in welding ordinary construction steels are rust and oil, which soil the welding edges, and also the moisture adsorbed by the surface of the electrode coating or of the flux grains. Another source of hydrogen may be moisture in the composition of a poorly annealed electrode coating or inside pumecious flux grains. There is no rust on the edges when chrome-nickel austenitic steels are welded. Hydrogen finds its way into the weld metal zone from the flux or from the coating. It was pointed out in Chapter 2 than in the manual welding of austenitic steels with base electrodes, the weld metal contains a minimum quantity of hydrogen. Oxygen-free fluxes, like low-silicon fluxes, introduce a considerable amount of hydrogen into the weld metal zone (see Table 16).

The solubility of hydrogen at room temperature in austenitic welds is considerably higher than in ferritic ones; the diffusion rate of hydrogen in austenitic welds is also very small, hence hydrogen which has found its way into the austenitic weld metal during welding is almost entirely retained by it for an indefinite time. In conjunction with Yu. V. Latash we carried out experiments to determine quantitatively the kinetics of the removal of hydrogen (at room temperature) from austenitic and ferritic welds (Table 39). From the experimental data given it is clear that in its initial state the austenitic weld does not contain very much more hydrogen than

than a ferritic weld, and that the quantity of hydrogen remains for practical purposes unchanged even after a lengthy period at room temperature. In the ferritic weld we observe an appreciable fall in the hydrogen content after five hours, and after 75 hours the hydrogen content has fallen to one sixth of what it was.

TABLE 39

Kinetics of Removal of Hydrogen from Weld Metal at +20°

Type of weld metal	Hydrogen content in weld metal in cm^3/100 g						
	Immediately after welding	Hours after welding					
		5	20	45	75	120	840
Ferritic weld in St. 3 steel	6.2	5.4	4.8	3.4	1.1	1.8	-
Pure austenitic weld in Kh23N18 steel	7.6	-	7.6	7.4	-	8.7	6.8

In 1950 we discovered that the type of current and its polarity have a considerable effect on the process of pore-formation in the welding of austenitic steels under fluxes containing SiO_2 and CaF_2.

The greatest porosity is observed when reversed polarity D-C is used for the welding. D. M. Rabkin has theoretically proved that in this case the weld pool is more effectively shielded from the hydrogen due to the formation of a layer of fluorine ions in welding under fluxes containing fluorine, or an excess of electrons in welding under fluorine-free fluxes /109/. In straight-polarity welding the number of pores is greater. The greatest porosity of all is observed in A-C welding (Fig. 84).

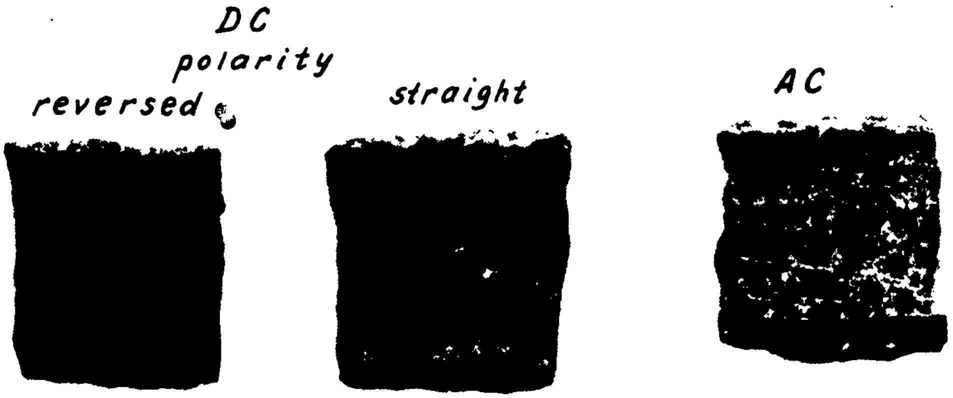

Fig. 84. Effect of type of current and polarity on the porosity of 18-8 steels welds (the spongy nature of the slag flake shows the tendency of the weld toward porosity .

The effect of the polarity of direct current on the hydrogen content in weld metal can be judged from the data in Table 40.

TABLE 40

Effect of Welding Current Polarity on Hydrogen Content and Porosity in Pure Austenitic Welds in 25-20 Steel.

Type of flux	Polarity	Hydrogen content in weld metal in $cm^3/100$ g	Pores present in weld
BKF-1 (5% SiO_2)	Reverse	7.8	No
	Straight	15.1	Yes
AN-26 (32% SiO_2)	Reverse	4.2	No
	Straight	6.0	No

The positive effect of D-C reverse polarity welding on reduction of porosity in welds makes it easier to use oxygen-free and low-silicon fluxes. It should be pointed out that manual welding with base-coated electrodes is also done with reverse polarity D-C.

The comparatively high pore-resistance of austenitic steels makes it possible to weld under conditions of high cooling rates of the weld (high rates, small cross-section welds), without fear of pores forming. This fact is of great practical importance since it is just these welding conditions which are the best for many austenitic steels.

As is known, in arc welding the metallurgical interaction of the slag, metal and gases is most violent in the core of the arc, where the highest temperatures are obtained, and the conditions are most favorable for contact between interacting phases. It is therefore not surprising that the hydrogen dissolves most in the drops of the electrode metal while the latter are in motion across the arc gap, and not in the weld pool. This is found in multi-layer welding. As each layer is reached, the amount of electrode metal in the weld increases together with the hydrogen content. Hence in multi-pass welding of austenitic steel porosity is usually found in the upper layers, while in single-pass welding porosity is comparatively rare.

It was pointed out in Chapter 2 that a higher arc voltage (longer arc) gives rise to an increase in the oxygen and nitrogen contents in the weld. This is not accidental. Multiple experiments in welding under different fluxes give the same result (Table 41).

Effect of Arc Voltage (Arc Length) on Hydrogen Content and Porosity in Pure Austenitic Welds

Type of steel	Type of wire	Type of flux	Welding conditions		Hydrogen content in weld metal in cm^3/100 g	Pores present
			Current in amps	Arc voltage		
Kh23N18 (EI417)	Sv-Kh25N20	BKF-1	440	28-30	13.2	Yes
		1% SiO$_2$	450	40-42	8.5	No
			450	42-44	7.0*	No
Kh23N23M3D3 (EI533)	Kh23N23M3D3 (EI533)	AN-23 20% SiO$_2$	550	30-32	4.8	No
		6.6% MnO	550	38-40	3.9	No

*The welding was carried out with a semi-shielded arc along the flux layer.

How is one to explain the reduction of the hydrogen content and the prevention of porosity in welding austenitic welds with a long arc? One of the reasons is that the gases apparently break through the covering of melted flux and escape from the fusion zone into the atmosphere. Another possible reason for the decrease in the hydrogen content in shielded-arc welding is an increase in the amount of melting flux. The more the flux melts and the more air finds its way into the fusion zone

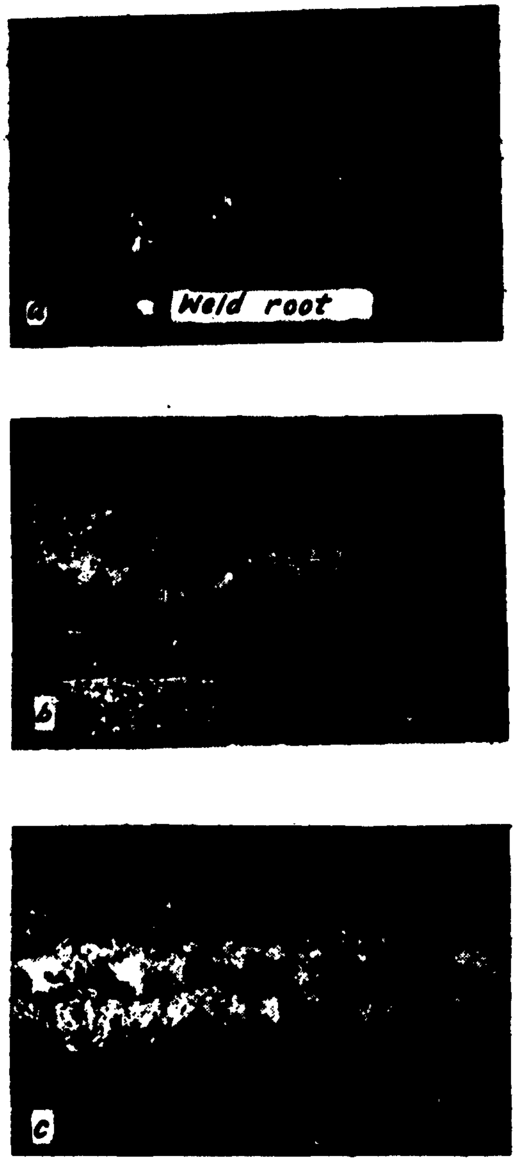

Fig. 85. Porosity of 18-8 steel welds due to hydrogen:

a) in the corner weld of a tee joint; b) in a butt weld;

c) porosity in a weld under AN-30 silicon-free flux.

from the spaces between the flux grains, the lower the partial pressure of the hydrogen and the smaller the amount of hydrogen reaching the weld pool, and the less the porosity. The data given in Table 41, as pointed out, show that the drop in the hydrogen content takes place at the same time as an increase in the oxygen and nitrogen concentrations in the weld. Unfortunately, it is not possible to recommend shielded-arc welding for austenitic steels. Although the danger of pores appearing is less, there is increased danger of the welds cracking on account of the oxidation of the ferritizers and possible austenitizing of the weld structure through the combined effect of the oxygen and nitrogen, and also as a result of an increase in the shape factor of the weld. It is essential, therefore, to find other means of reducing the hydrogen content in the weld metal when welding austenitic steels. This is especially necessary since austenitic-ferritic welds, which are given preference in view of their high resistance to hot cracks and intergranular corrosion, are much more liable to form pores than pure austenitic steels (Fig. 85). This fact may be connected with a greater decrease in the solubility of hydrogen when crystals of the δ-phase are present in the crystallizing weld.

In welding low-carbon and low-silicon steels the surest method of preventing hydrogen porosity is to use high-silicon fluxes containing calcium fluoride.

In submerged-arc welding with these fluxes, as shown in the work of I. V. Kirdo and V. V. Podgayetskiy, the following reaction takes place in the gas phase

$$SiO_2 + 2CaF_2 + H \rightleftharpoons 2CaO + SiF + 3HF. \quad (10)$$ [*]

As a result of the combining of the hydrogen into HF, which is insoluble in the weld metal, the hydrogen content decreases and porosity is averted.

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[*] Translator's note: Sic! The equation is not balanced.

V. V. Podgayetskiy and D. M. Rabkin have shown in their work that prevention of porosity in welding conventional steels is also possible if the hydrogen is oxidized into the hydroxyl OH, which, like HF, is insoluble in the liquid metal. This method of using oxygen to combat hydrogen porosity is not practised in welding carbon steels, but has been used successfully in automatic semi-shielded-arc welding of aluminum and also in argon-arc welding.

Experiments which we carried out in collaboration with Yu. V. Latash showed that in welding under oxygen-free flux consisting mainly of CaF_2 and NaF and not containing SiO_2, the hydrogen may combine into CaF and F as a result of the dissociation of CaF_2. This process, however, is rather feeble and for this reason if the silicon concentration is small, pores may form. At the same time, even a small amount of TiO_2 added to an oxygen-free flux makes possible a sharp reduction in the hydrogen content of the weld. For example, Kh23N23M3D3 (E1533) steel was welded by the multipass method with the conditions remaining the same each time; in the case of the BKF-1 oxygen-free flux the upper layer contained 9.3 cm^3/100 g hydrogen and was damaged by pores. When the same flux was used with the addition of 10 to 15% TiO_2, the hydrogen content dropped to 5.4 cm^3/100 g, and there were no pores.

It is of interest to determine the mechanism of the positive effect of adding TiO_2 to the flux. For this we will use the following thermodynamic equations:

$$TiO_2 + H \rightleftarrows \tfrac{1}{2}Ti_2O_3 + \tfrac{1}{2}H_2O \qquad (11)$$

The change in free energy of the system in standard conditions can be calculated from the equation

$$F^0 = \Delta H^0 - T\Delta S, \qquad (12)$$

where $H^0 = H^0_2 - H^0_1 = -Q_p$ is the heat effect of the reaction or the difference in the heats of formation of the final (H^0_2) and initial (H^0_1) substances; $S^0 = S^0_2 - S^0_2$ or the difference in entropies in the final (S^0_2) and initial (S^0_1) substances.

Substituting the corresponding values of the constants /29/, we obtain for the reaction (11):

$$\Delta F^0 = -49300 + 7.4T$$

The nature of the variation in free energy ΔF^0 with temperature in the given reaction is illustrated by the graph in Fig. 86.

It can be assumed that the hydrogen is oxidized by another reaction as well:

$$2TiO_2 + H \rightleftarrows Ti_2O_3 + OH, \qquad (13)$$

in this case $\Delta F^0 = 4670 - 11.33\,T$.

Comparison of the values of the free energy obtained with the corresponding data for the known reaction (10) show the possibility in principle of the hydrogen being oxidized by the oxygen in the titanium dioxide.

It is natural to assume that the prevention of porosity in austenitic welds in welding under fluxes containing fluorite (CaF_2) and TiO_2 may occur as a result of the hydrogen combining with the fluorine. Here the fluorine may be carried to the hydrogen by the TiF_4 in similar fashion to what occurs when silicon tetrafluoride is present:

$$TiO_2 + 2CaF_2 \rightleftarrows 2CaO + TiF_4 \qquad (14)$$

and further

$$TiF_4 + 3H \rightleftarrows TiF + 3HF. \qquad (15)$$

The overall reaction involving combination of the hydrogen by means of titanium fluoride will be as follows:

$$TiO_2 + 2CaF_2 + 3H \rightleftarrows 2CaO + TiF_4 + 3HF. \qquad (16)$$

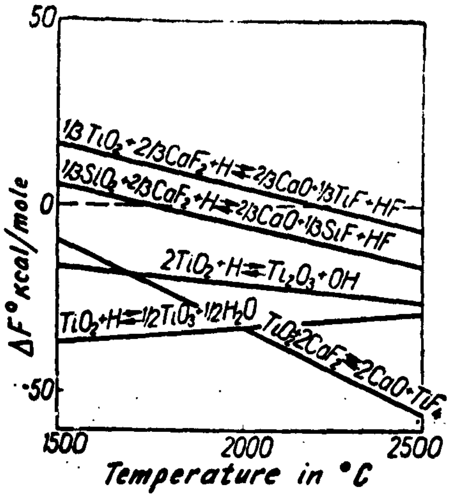

Fig. 86. Variation in free energy ΔF^0 with temperature for different reactions in which hydrogen combines into HF and OH.

To solve the problem of the liklihood of reactions (14) and (15) occurring under the welding conditions, it is essential to work out the heat of formation and the entropy of the gaseous TiF_4 and TiF, since these data are not given in any of the works with which we are acquainted. A description of the properties of the two fluorides TiF_4 and TiF is contained in the reference work, Gmelins Handbuch. It points out that the spectrum of gaseous TiF is found between 1000 and 2000°C (in the region of 2000 to 5000 Å).

To determine the heat of formation H°_{298} of TiF_4, we apply the graph method proposed by A. F. Kapustinskiy. The initial condition for the solution is a linear dependence between the heats of formations ΔH of the chemical compounds referred to a single equivalent ($-\frac{\Delta H}{\underline{W}}$ where W is the valency) and the logarithm of the element numbers log $\underline{Z}$, which are electronic analogs, i.e., they make up one group in Mendeleyev's periodic table:

$$-\frac{\Delta \underline{H}}{\underline{W}} = \underline{K} \log \underline{Z} \tag{17}$$

In the case of titanium the analogs, whose thermal constants are known (we are concerned with tetrafluorides), are carbon and silicon, and calcium and potassium for titanium fluoride.

Using a graph we find $H^0_{298} = 424$ kcal/mole for $TiF_{4\ gas}$, and $H^0_{298} = 33.5$ kcal/mole for TiF_{gas}.

To determine the entropy of TiF_4 we can use the graphic method put forward by V. A. Kireyev and developed by M. Kh. Karapet'yanets.

There is a linear dependence between log S and log M (where S is the entropy, and M the molecular weight) for similar compounds (for example, fluorides) of elements in the same group or same series of the periodic table. We find that S^{o}_{298} is equal to 74.4 $\frac{\text{cal}}{\text{deg mole}}$ for TiF_4. We calculate the entropy of TiF from the empirical formula put forward by O. Kubashevskiy and E. Evans. It equals 54.87 $\frac{\text{cal}}{\text{deg mole}}$.

Having the values H^{o}_{248} and S^{o}_{248} for TiF_4, we find that the variation in free energy for reaction (14) during the formation of TiF_4 obeys the law:

$$\Delta F^{0} = 47600 - 48.4\,T,$$

and for the reaction in which hydrogen combines into HF (15)

$$\Delta F^{0} = 26100 - 7.6\,T.$$

For the total reaction (16)

$$\Delta F^{0} = 60500 - 23.2\,T.$$

The relevant data are given in Fig. 86.

And so it can be seen on the basis of thermodynamic calculations that the reduction in the hydrogen content and prevention of porosity in austenitic welds under a flux consisting of CaF_2 and TiO_2 is due to the hydrogen combining into the insoluble (?) compounds OH and HF. The preponderant development of the one or the other reaction in which the hydrogen combines apparently depends on many factors. Nevertheless, the thermodynamic calculations fully support the cases in which the addition of TiO_2 has a favorable effect.

In a number of instances, for example, when the ferrite concentration in the weld is high, TiO_2 may be added to the flux to avoid porosity without interfering with the resistance of the weld to hot cracks. But with pure austenitic welds the addition of TiO_2, in the same way as SiO_2, is inadmissible since the hot brittleness increases.

In welding austenitic steels a very effective way of reducing the hydrogen content and preventing porosity is to add a small quantity of oxygen to the fusion zone, but not in the form of the oxides SiO_2 and TiO_2, or manganese ore containing phosphorus, as is done in welding ordinary steels. It is best to add the oxygen in the form of pure manganese oxides. An idea of the effect of adding oxygen in the form of $KMnO$ on the amount of hydrogen in an austenitic weld can be gained from the data in Table 42. A sure way to reduce the hydrogen content in austenitic welds is to use carbon dioxide as a preservative.

TABLE 42

Effect of Adding $KMnO_4$ to Oxygen-Free Flux on Gas Content in Austenitic Kh23N18 Steel Weld.

Addition of $KMnO_4$ in g/100 lin. mm of weld	Gas content in weld		
	Hydrogen in $cm^3/100$ g	Oxygen in %	Nitrogen in %
None	7.8	0.0289	0.0322
1	7.2	0.0392	0.0387
3	5.5	0.0573	0.0296
6	3.6	0.0800	0.0151

The more oxygen there is in the austenitic weld, the less ~~the~~ hydrogen. This function is clearly shown by the graph in Fig. 83 b. Exactly the same relationship was found by Reeve in the USA and V. V. Bazhenov /16/ with respect to ordinary carbon steel welds.

Thus, the less oxygen in the austenitic weld, the more hydrogen there is in it. Aluminum, titanium, silicon and other ferritizers, and also manganese, being strong oxidizers, sharply decrease the oxygen content in the liqid pool in welding austenitic steels, the inevitable result of which is an increase in the hydrogen content. This is the reason for the great tendency of austenitic-ferritic steels to form pores compared with pure austenitic steels. Another reason for this increase when the silicon, titanium and alumimum content of the weld is high is the formation of ferritic regions where the solubility of hydrogen is less than in austenite.

It ensues from what has been said in this section that in austenitic steel arc welding one need only beware of the porosity caused by hydrogen. Removal of the chief sources of the hydrogen -- moisture, rust, oil and other organic forms of contamination -- and the use of reversed-polarity D-C rid us of this danger to a considerable extent. In welding with oxygen-free and low-silicon fluxes, a new and reliable method of preventing porosity is to add small quantities of oxygen to the flux, for instance, in the form of the higher oxides of manganese, chromium or iron.

In austenitic-ferritic welds possessing a high ferrite concentration and great resistance to hot cracks, TiO_2 can be added to oxygen-free fluxes for the same purpose.

In a case where the hydrogen has difficulty in escaping from the crystallizing weld pool, the surface of the weld may form depressions and appear "worn out", although the weld itself contains no pores. This occurs in welding under fluxes which form short slags, the viscosity of which at the solidification point of the weld is so great that the hydrogen, unable to pass through the thickening slag, has to move over the interface between the metal and slag, leaving traces on the weld. All that needs to be done to avoid this state is to decrease the viscosity of the slag at the solidification point of the weld, even if the amount of hydrogen added to the weld pool is the same as in welding under a viscous flux. To reduce the viscosity of a CaF_2-base flux we recommend the addition of 5 to 25% NaF (ANF-5 flux). In machine-arc welding high-nickel Kh20N80 alloys the addition of NaF to an oxygen-free flux has a striking effect and is a reliable way of getting rid of the "worn out" effect.

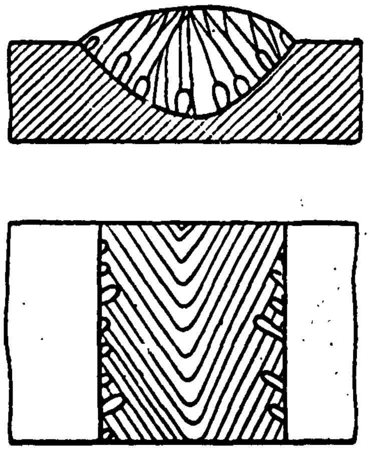

Fig. 87. Pores along the weld due to hydrogen which has found its way into the weld from the base metal.

In conclusion mention should be made of another type of porosity in welds which occurs when metal with an excessively high hydrogen content is welded. As is known, nickel when crystallizing dissolves twice as much hydrogen as austenitic steel. Electrolytic steel contains an especially large amount of hydrogen. In welding nickel, when the arc melts the butts being welded, hydrogen enters the weld in great quantities. Because of the high crystallization rate of the peripheral regions of the weld, the hydrogen does not have time to escape, but reacts with the nickel oxide in the following way

$$NiO + H_2 \longrightarrow Ni + H_2O.$$

Pores form along the line of fusion in the process (Fig. 87).

Pores are also found along the line of fusion in argon-arc welding of austenitic-ferritic 17-7RN steel /169/.

In the argon-arc welding of nickel a sure remedy for preventing porosity is hydrogen. If argon and hydrogen are used as a reducing atmosphere, the formation of nickel oxide is impossible, and, consequnetly, the appearance of the hydrogen defect.

In the argon-arc welding of 17-7RN austenitic steel an apparent reason for porisity is the hydrogen with which the base metal becomes saturated during electrolytic etching. The addition of oxygen to the argon would probably help in this case.

Another possible cause of porosity along the fusion line in the welding of austenitic steels with an increased aluminum content (17-7RN steel contains more than 1% aluminum) may be local generation of great amounts of nitrogen due to the decomposition of the aluminum nitride. To avoid the porosity caused by the nitrogen, small amounts of titanium and zirconium are added to the composition of 17-7RN steel. These elements bind the nitrogen into stable nitrides which are still preserved at the melting point of steel.

CHAPTER 5

CORROSION RESISTANCE OF THE WELD JOINTS OF CHROMIUM-NICKEL AUSTENITIC STEELS

A great majority of the chromium-nickel austenitic steels are used as acid resistant materials for work in aggressive liquid media, and therefore they must have a high corrosion resistance. As a result of many years of research by Soviet metallurgists there were created dozens of makes of austenitic steels which excellently serve under complex conditions in chemical, petroleum and other branches of industry. Heat-resistant steels which work in aggressive gaseous media must ~~have~~ be ~~resistance to scaling~~ scale-resistant.

The problem of ensuring the required corrosion resistance of weld joints in different media and at different temperatures is of great practical importance and, together with the problem of increasing the resistance of welds against the formation of hot cracks, is a deciding factor in the whole problem of welding of austenitic steels. Therefore, particular attention is paid to the problem of the corrosion resistance of weld joints.

I. Classification of Kinds of Corrosion Destruction of Weld Joints

At the present time two basic types of corrosion destruction of the weld joints are known: intergranuler (or structural*) corrosion and over-all corrosion. Fig. 88

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* The term "structural corrosion" was proposed by G. V. Akimov /4/.

schematically represents both kinds of corrosion destruction. Liquid and gas corrosion are differentiated.

Intergranular liquid corrosion can develop in three sections of the weld joint:

a) in the base metal at some distance from the weld;

b) in the weld metal;

c) in the base metal immediately at the fusion line (so-called knife-line corrosion) Fig. 88c.

The mechanism of weld-joint destruction which results from structural corrosion, is in general the same and does not depend on the location -- whether it is the weld or the base metal. The structural corrosion results from the penetration of an aggressive reagent into the stainless steel along the grain (crystal) boundaries, and for this reason this kind of deterioration was called intergranular corrosion. In structural corrosion the individual grains are disconnected and the metal disintergrates into a powder. The extent of the intergranular corrosion cannot be determined from the outer appearance of the metal because the penetration of the aggressive medium into steel occurs along the austenitic grain boundaries, and the grains (crystals) themselves remain untouched. To determine the extent of corrosion it is necessary to load the metal -- bend it in such a way that the affected area is in the zone under tension, or stretch it.

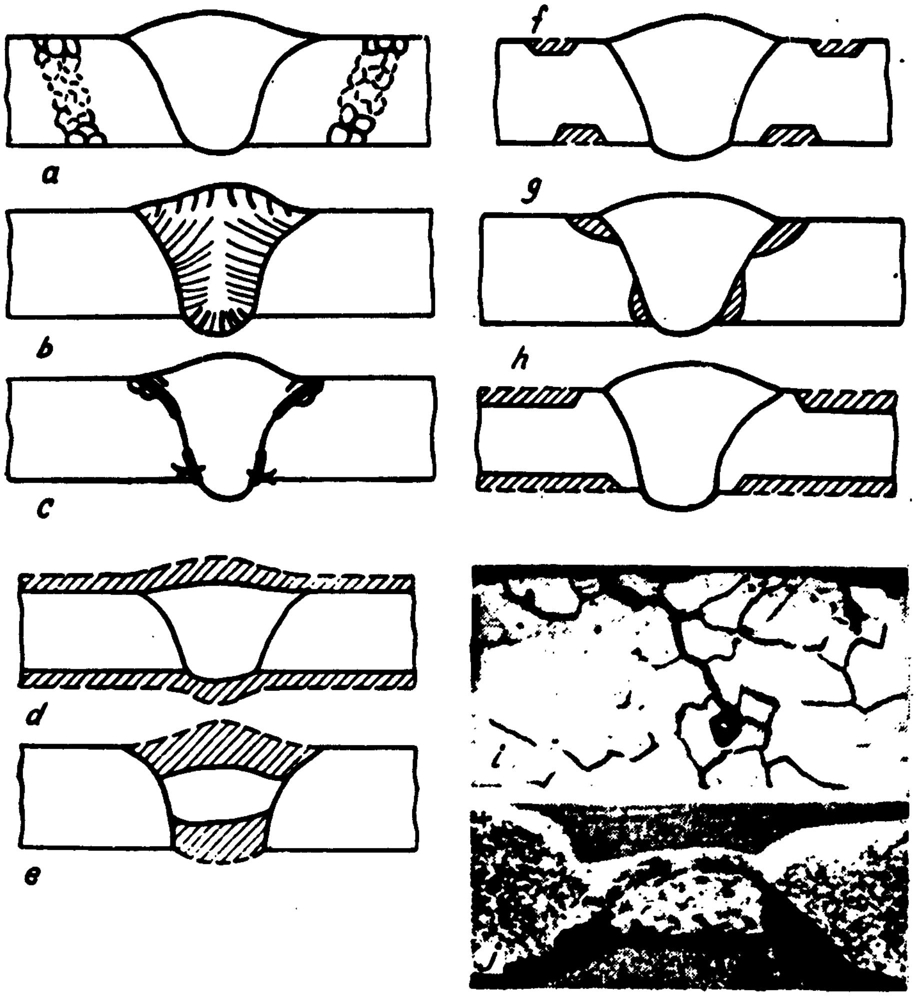

Fig. 88. Types of weld joint corrosion deterioration:

Intergranular (structural) corrosion: a) in the base metal; b) in the weld metal; c) knife-line corrosion along the fusion line.

General corrosion: d) uniform; e) concentrated on the weld; f) same on the base metal; g) knife-line at the fusion line; h) accelerated corrosion of the base metal; i) intergranular corrosion in 1Kh18N9 steel, caused by heating during welding (magnified 300 times); j) concentrated corrosion on the weld.

The corrosional cracks will open and the individual austenite grains may fall out from the deformed metal. Because of the disconnection between grains, the affected metal loses its metallic sound. A sample which is deeply affected by the intergranular corrosion has a dead ("cardboard") sound when it hits metal or stone (concrete, marble). Weld joint deterioration caused by the development of intergranular corrosion is illustrated by the photographs given in Fig. 89.

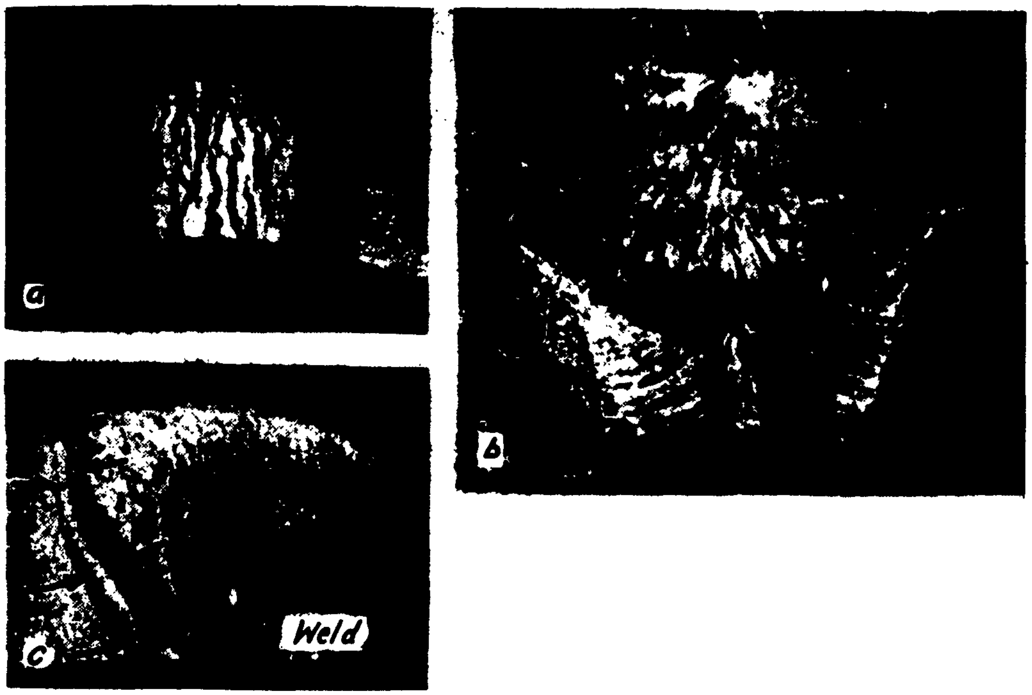

Fig. 89. Examples of the weld joint intergranular corrosion:

a) single-pass weld, b) multiple-layer weld, c) concentrated intergranular corrosion along the boundary between the weld and the base metal (knife-line corrosion).

Though the mechanisms of intergranular corrosion in the weld and in the base metal are the same, the causes of this type of corrosion are different at different weld joint localities and depend on many factors. We will deal only with the most important ones.

The appearance of <u>intergranular corrosion at a distance from the weld</u> is caused by effect of the welding heat on the steel. Along both sides of the weld there are areas of base metal which as a result of welding were subjected to a more of less prolonged heati at critical temperatures which cause decomposition of the austenite and precipitation of the excess phase along the grain boundaries. It is thought that the precipitation of the excess phase is accompanied by a decrease in chromium concentration at the surfaces of t austenite grains. If chromium content in the boundary layers of austenitic grains (cryst

falls below the threshold of [illegible]),
the aggressive agent dissolves the chromium-poor metal and starts intergranular corrosion. Steels which are susceptible to intergranular corrosion as a result of short-term heating in the region of critical temperatures are not fit for welding and at the present time find only limited uses. Steels which are stabilized with corresponding amounts of titanium or niobium and tantalum are stable against critical temperatures and are therefore widely used in industry. Because of this, when welded structures are made from stainless steels one usually does not have to worry about the intergranular corrosion at some distance from the weld if the quality of the base metal was properly controlled.

Intergranular corrosion in the weld metal can occur for two reasons: 1) the weld in the natural state, i. e., after welding, was liable to intergranular corrosion because of precipitaion in it of an excess phase; 2) the weld was corrosion resistant in its natural state, but lost the resistance because of the influence of critical temperatures during its service.

The intergranular corrosion in the base metal next to the fusion line, or knife-line corrosion, affects a narrow strip of steel which, as a result of the welding heat cycle, was heated to temperatures exceeding 1250°C. This type of corrosion affects only steels which are stabilized with titanium or niobium and tantalum. If the steels are heated to temperatures exceeding 1200 to 1250°, carbides of titanium or niobium dissolve in the austenite. During the subsequent influence of critical temperatures at the section of the base metal which were heated to the temperature of carbide dissolution, titanium and niobium remain in the solid solution, the carbides of chromium precipitate along the grain boundaries and intergranular corrosion starts to develop. Because of the high heat concentration during arc welding, the superheat area next to the weld is very narrow and therefore the corrosion destruction has a concentrated character. Usually a strip of the base metal up to 1 to 1.5 mm wide is affected. If the samples affected by this type of corrosion are tested by bending, the destruction appears as a knife cut, which resulted in the name "knife-line corrosion."

It must be emphasized that this type of intergranular corrosion is observed in steels which in finished form are insensitive to critical temperatures.

The over-all liquid corrosion causes a visible destruction of the weld joint and is a result of metal dissolving in the aggressive medium.

The general corrosion can also result from intense destruction of the grain boundaries, i. e., intergranular corrosion. As a result of the bonding loss among the crystals individual austenitic grains can be carried away by the aggressive liquid. The steel grains themselves in this case can have sufficient stability in the given aggressive medium.

In over-all corrosion there is most often a dissolution of both the austenitic grains and the boundaries between them.

The over-all corrosion can be uniform or concentrated.

In the first case the corrosion rates are approximately the same for all the sections of the weld joint --the base metal, the weld, and the zone next to the weld. If the corrosion develops uniformly, it is not dangerous because the rate of metal deterioration can be calculated and the structure life can be determined beforehand.

The concentration of over-all corrosion on the weld or in the zone next to it is very dangerous. The concentration of the over-all corrosion can take place at some distance from the weld in the area where the steel is affected by critical temperatures and at the fusion line, i.e., where the intergranular knife-line corrosion develops. In concentrated over-all corrosion the absolute metal loss which results from its dissolution is relatively small. But the rate of the metal thinning at the concentration area is tens and hundreds fold higher than the thinning rate of the steel from over-all corrosion in the given medium. This can lead to a very fast destruction of the welded apparatus. Thus, for example, in England a welded tank 6 m in diameter, 6 m high and 2.6 mm thick, which was made of 18-8 steel with titanium and used for storage of 96% nitric acid, went out of service after thre years of operation because of knife-line corrosion /208/. There are cases when apparatus made of 1Kh18N9T steel failed after a few weeks of service because of concentrated corrosion of the welds. In the USA a container made of 18-8 steel with niobium was destroyed extremely fast as a result of knife-line corrosion /174/.

It must be emphasized that in spite of the undoubted connection between the processes of over-all and intergranular corrosions, a stabilization of the metal against one of the types of destruction does not guarantee a sufficient stability against the other type of corrosion. Thus, for example, a weld may be stable against intergranular corrosion in nitric acid, but at the same time it may be intensely attacked by over-all corrosion in the same medium.

There is one more type of liquid corrosion which occurs as a result of the combined influence of tensile stresses and aggressive medium: it is called the corrosion under stress /23/. The metal destruction from the corrosion under stress has a mixed character -- both intergranular and transcrystalline. It is known that the corrosion under stress usually develops when halogen salts are present in the electrolyte, though it may occur in other media -- for example, in the molten eutectoid lead-bismuth at 400--500°.*

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* Newest investigations showed that stable austenitic steels and welds are liable to transcrystalline corrosion under stress in sulfuric acid /77/.

As was established in the Institute for Electric Welding /77/, some austenitic steels are subject to intergranular corrosion as a result of cold hardening (see Section 10). The mechanism of corrosio under stress is still insufficiently explored.

Finally, in weld joints of austenitic steels, intergranular corrosion may be caused by fatigue. The development of this type of corrosion is greatly influenced by the effect of the adsorption weakening of a metal, known as the Rebinder effect /112/.

All that we said about corrosion refers to the performance of weld joints in liquid aggressive media. However, welds of high-nickel scale-resistant steels, such as type 25-20, are as a rule not require to perform in liquid media, but in gas, where they are exposed to the effect of the so-called gas corrosion.

Gas corrosion may be of an intergranular character. This type of destruction is usually observed on high-nickel steels, alloys and welds performing in an atmosphere of sulfurous gases (see Fig. 115). Over-all corrosion, or scale formation, can be either uniform or concentrated on welds if these contain excessive amounts of vanadium or other elements impairing their resistance to scaling (see Fig. 114).

It is the liquid corrosion of welds that is essentially discussed in this volume.

2. Modern Theory of the Intergranular Corrosion of chromium-nickel austenitic steels

Nonstabilized chromium-nickel austenitic steels (~~that means~~ those not alloyed with titanium, niobium or tantalum) of normal carbon content (above 0.02 to 0.03 per cent) become susceptible to intergranular corrosion as a result of being slowly cooled from temperatu higher than 1000°C, or due to holding for longer or shorter periods of time in the so-called critical temperature range of 450 to 850°C (according to other data from 400 to 900°C). We also know that the phenomenon of ~~of~~ intergranular corrosion of stainless steels is linked to the precipitation of chromium-rich iron~~-~~ and chromium carbides, $(Cr,Fe)_4C$ (or $(Cr,Fe)_{23}C_6$, according to other data) along the boundaries of austenite grains. According to the pseudobinary diagram of state (Fig. 2), the maximum solubility of carbon in austenite at normal temperatures is 0.02 to 0.03 per cent.

Hence, in 18-8 steel, precipitation of carbides can occur only at a higher carbon content. The formation of these carbides proceeds at the expense of large amounts of chromium being drawn from the adjoining peripherial regions of the austenite grains. This loss in chromium impairs the chemical stability of the boundary layers of the austenite; as a result, the steel becomes susceptible to structural corrosion.

The modern theory of intergranular corrosion is based on the widely known theory of multielectrode systems which was developed by the prominent Soviet scientist G. V. Akimov and his associates /4/,

A. N. Frumkin /135/ and others. This theory regards stainless steel susceptible to intergranular corrosion as a system of the following three electrodes (Fig. 90): I) the austenite grain; II) chromium and iron carbides at the boundaries of grains of austenite; III) chromium-impoverished boudary layers of austenite. Under the effect of oxygen in an aggressive medium, grains of austenite containing a sufficient amount of chromium and chromium-rich carbides will passivate and acquire a positive potential. However, the boundary layers of the

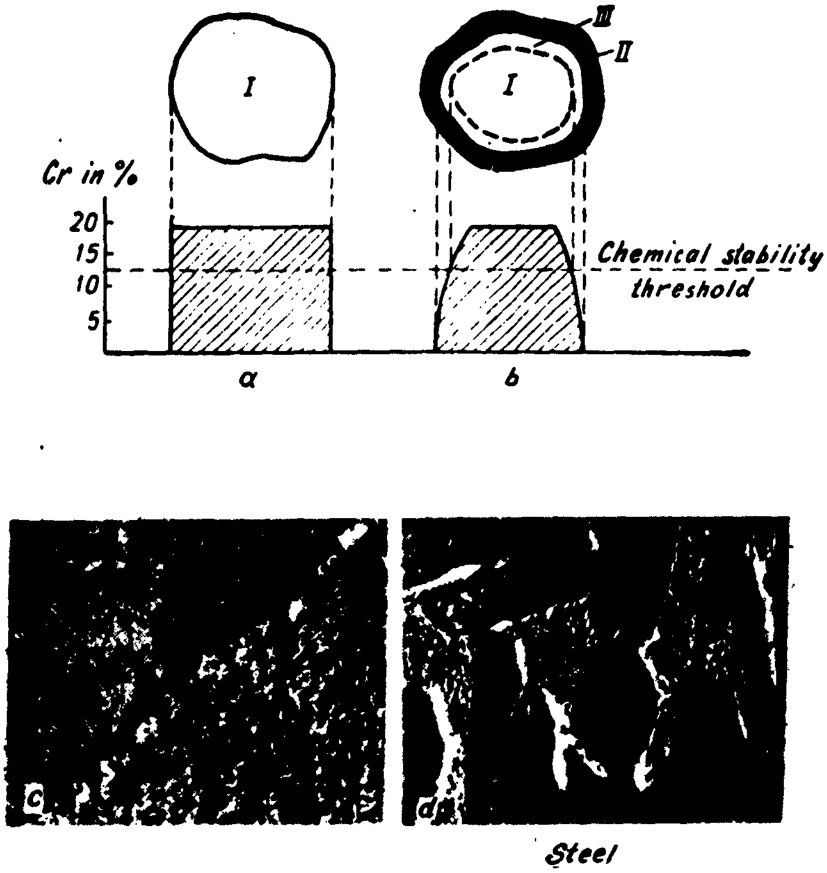

Fig. 90. Three-electrode system:

a) hardened 18-8 steel, b) 18-8 steel after critical temperatures: I) austenite, II) Cr and Fe carbides at the grain boundaries; III) chromium-impoverished boundary regions (A. P. Gulyayev), c) chromium- and iron carbides at the grain boundaries of 18-8 steel, (X 12500) (Yu. B. Malevskiy), d) carbides in the ferrite regions of a ferrite-austenitic weld of 18-8 steel (X 6000) (Yu. B. Malevskiy) austenite grains which have lost large amounts of chromium to the carbides cannot passivate, assuming thus a negative potential. In the three-electrode system under discussion, the electrodes I and II (grain and carbides) are the cathodes with respect to the anode III austenite boundary layers). Only a relatively small surface of the austenite anode boundary layers is exposed to the effect exerted by a large surface of the cathode grains, as a result of which the corrosion current reaches a large magnitude, while intergranular corrosion develops actively and swiftly.

The sensitivity to intergranular corrosion of austenitic chromium-nickel steels depends on the time of their exposure to a given critical temperature. The time in which the steel preserves its initial resistance to intergranular corrosion (we will denote it by T_1) varies with the critical temperature. Curve 1, in Fig. 91, illustrates the interdependence of the annealing temperature and time of 18-8-steel welds, as well as their susceptibility to intergranular corrosion. N. Yu. Pal'chuk obtained analogous curves for manual welds /96/.

The curves under discussion show a drop in time T_1 from several hours to mere minutes with annealing temperature rising up to 650 to 700°C. Susceptibility to intergranular corrosion is absent altogether as soon as temperature reaches 800 to 850°C. Behavio of steel varying with temperature is explained by the different rates of two opposite processes, namely, precipitation of chromium carbides, and chromium diffusion from the central region of the austenite grains to their periphery. The rates of austenite disintegration and chromium-carbide formation increase with temperature rising from 450 to 500°C up to 650 to 700°C more rapidly than the rate of chromium diffusion. In view of this, the steel acquires susceptibility to intergranular corrosion in less time.

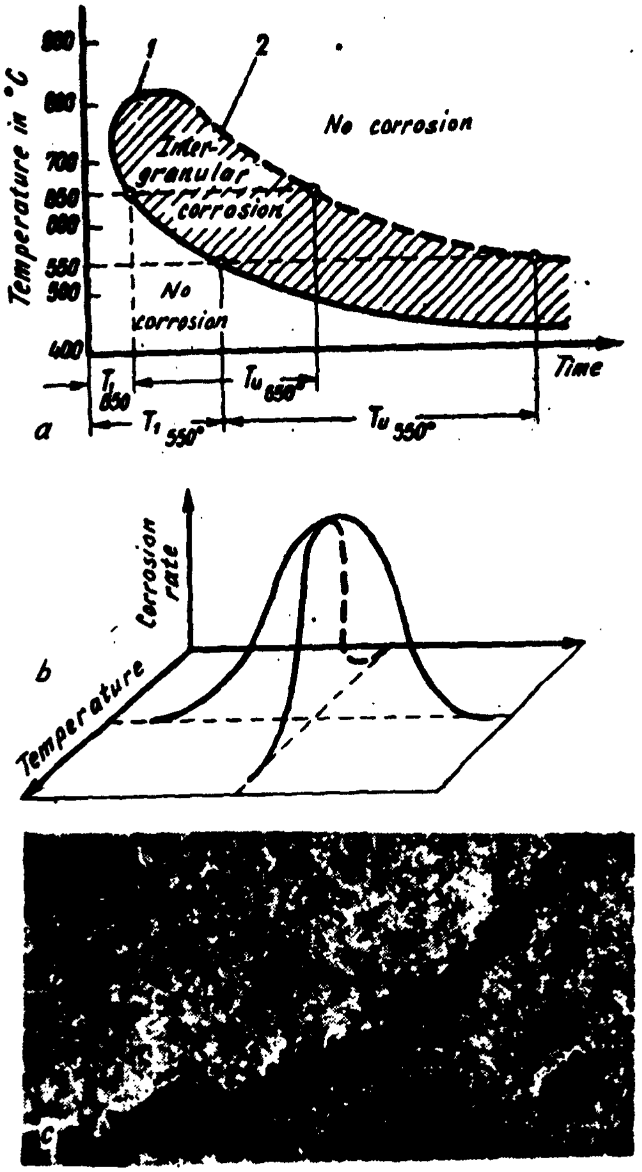

Fig. 91. Effect of holding time and temperature on the susceptibility 18-8-steel welds to intergranular corrosion (diagram): a) according to data of this author and those of Zitter and others /266, 264/, b) according to Seferian, c) carbides at the austenite grain boundaries in 18-8-steel weld (X 2500) (Yu. B. Malevskiy).

A further temperature rise from 650 to 700°C up to 800 to 850°C leads to a steadily increasing rate of chromium diffusion. It is assumed ~~presumed~~ that chromium-carbide precipitation is accompanied by an ever growing rate of chromium-atom displacement from the more distant to the peripheral regions of austenite in which, latter, the concentration of this element is temporarily reduced. In other words, the chromium consumed in the formation of carbides is supplied also by the more distant regions of the austenite, not only by the peripheral ones. In view of this a dangerous local impoverishment of austenite in chromium does not occur as rapidly as at lower temperatures, where diffusion from within the grains is too slow to compensate chromium consumption out of the peripheral zones. With rising temperature, moreover, the process of coagulation of the carbides and their disolving in the austenite begins to tell.

It was established that long holding of 18-8 steel in the critical temperature range has the effect of rendering the metal quite insusceptible to intergranular corrosion. According to data of Bain and others /147 and 148/, 0.08% carbon-bearing 18-8 steel becomes insensitive to intergranular corrosion after having been held for thousands of hours at a temperature of 650°C. This phenomenon is explained by chromium diffusion from within the austenite grains toward their boundaries, and by a higher chromium concentration in the regions previously impoverished in chromium. From the above it follows that apart from the first period of holding 18-8 steel at critical temperatures (that is, the period T_1)

there exists also another period, which we have named the period of immunization, T_{im}. Thus, 18-8 steel exposed to the effect of critical temperatures, will at first retain its resistance to intergranular corrosion for a certain time T_1. After a more or less lengthy exposure to the given temperature, the steel will regain its temporarily lost former resistance. Considering what we have said above, it seems that another curve besides curve 1 may be plotted on Fig. 91, a, namely curve 2, expressing the variations of the immunization period T_{im} as a function of temperature.

As has been stated earlier, the immunization period of rolled steel lasts hundreds and even thousands of hours. In view of this, the problems related to the immunization of rolled 18-8 steel, i.e., the recovery of the resistance which had been temporarily lost, or the so-called secondary resistance of the steel,* have no practical value. It will be seen below that immunization heat-treatment of welds yields positive results in incomparably shorter periods of time, and thus merits particular attention (see Section 6).

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* The definition "secondary resistance" was proposed by N. Yu. Pal'chuk /98/.

Fig. 91, b schematically illustrates the variation character of the corrosion resistance of 18-8 steel as a function of annealing temperature and holding time.

The theory of intergranular corrosion of stainless steels, briefly outlined above and put forward by Schottky /232/ and Bain with associates /147/ is based on the recognition of the fact that the formation of chromium carbides is accompanied by loss in chromium in the boundary layers of the austenite grains (crystals), hence this theory is also called the "theory of impoverishment" or the "carbide theory".

It is of interest to note that in 1931, when Bain first introduced his theory of impoverishment, the world still lacked the modern devices permitting observation of the chromium carbides at the grain boundaries of austenitic steel. In this Bain succeeded 25 years later using an electron microscope /149/ *, the carbides being discovered directly as they dislocate along the grain boundaries of the austenite and not in their extracted form as seen by Mala and Nielson,and G. G. Mukhin /83/. Though the presence of carbides at the grain boundaries is now proved, we still lack convincing experimental evidence on the impoverishment of austenite-grain boundaries in chromium.

At the present time, the theory of impoverishment is subject to severe critism despite it being supported by an overwhelming majority of corrosion experts. In his work, I. A. Levin /59/ comes

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* Analogous investigations were carried out on our request by Yu. B. Malevskiy in the Institute for electric welding. (See Fig. 91, /246/.

to the conclusion that this theory is incompetent, pointing out that destruction of steels due to intergranular corrosion proceeds directly along the grain boundaries where, apparently, chemically unstable iron carbides are precipitated, and not along the peripheral regions of the chromium-depleted grains. According to I. A. Levin /60/, the reason for the localization of corrosion at the grain boundaries is to be seen in the great mechanical stresses caused by carbide precipitation. In another work /173/ which also questions the correctness of the said theory, it is suggested that the reason for corrosion destruction should be sought in the precipitation of carbides rich in iron rather than in chromium, and therefore lacking chemical stability. Here, in essence, it is spoken of the destruction of the intercrystalline layers proper, and not of the peripheral regions of austenite grains.

In the literature we find numerous examples for the intergranular corrosion of stainless steels which cannot be expleined on the basis of the carbide theory; such as:

1) Vacuum-smelted stainless steel bearing 17.6 per cent chromium and 24.0 per cent nickel while, practically, carbon-free (a mere

0.003% C), is susceptible to intergranular corrosion. It became susceptible to corrosion after holding for 169 hrs at 500°C.

2) Molybdenum-bearing 18-8 steel containing 0.017% carbon, is susceptible to intergranular corrosion in 65-per cent nitric acid after prolonged exposure to critical temperatures.

3) Stainless steel,containing 0.03% carbon, 18.12% chromium and 8.34% nickel, proved susceptible to intergranular corrosion after a holding of 24 hrs at 650°C and boiling in a sulfuric-acid solution of copper sulfate for 144 hrs.

4) Stainless steel, containing 17% chromium, 7% nickel, 0.7% carbon, and completely free of (the energetic carbide-formers) titanium and niobium, is entirely immune to intergranular corrosion (in the region of the weld) under the effect of welding.

In the first three cases we are dealing with intercrystallitic corrosion in steels with an extremely low carbon content. It is unlikely that with a 0.003% carbon content in the steel corrosion would be caused by chromium carbides. Carbide precipitation is not to be expected with such a low carbon concentration. In the last case (fourth example) there was no corrosion, although there was nothing to prevent carbide precipitation. Aluminum does not produce carbides and the steel does not contain any carbide-forming elements stronger than chromium. What prevented the appearance of corrosion ? The carbide theory cannot provide an answer to this question.

The nature of intercrystallitic corrosion of stainless steel is very complex and multiform. The appearance of this kind of corrosion in welds is related to the existence of many additional factors, including internal stresses which facilitate the appearance of Rebinder's "adsorption-wedging" effect, thus making it difficult to detect the real causes of the structure of corrosion.

Certain authors assumed that the cause of intercrystallitic corrosion lies in the oxides and nitrides that precipitate on the grain boundaries /222/. It has been assumed that the σ-phase may be the cause of this kind of corrosion. We advanced the hypothesis that intercrystallitic corrosion in 18-8-type steel welds is caused by an impoverishment of the austenite grain boundaries, due to the precipitation of secondary chromous ferrite, instead of chromium carbides /63/.

In order to explain the causes of intercrystallitic corrosion, Althof advanced the so-called intra-atomic theory, which, however, did not receive acceptance.

It has been advanced that intercrystallitic corrosion is caused by the formation -- on the grain boundaries -- of a troostite structure which does not resist the action of an agressive medium (Kinzel), and also the formation of a stressed state created by an excess phase on the grain boundaries /60/.

None of these hypotheses, including the carbide theory, can be considered sufficiently founded.

Henceforth, when discussing problems of corrosion in welds, we will proceed from the theory of impoverishment, assuming, however, that austenite is being impoverished of chromium not only because of chromium carbides, but because of an excess phase produced by the disintegration of austenite, which phase is a complex of structural formations, i. e., of carbides and secondary ferrite /98/.

Section 3. The Effect of Primary Ferrite on the Corrosion Resistance of Austenitic Chromium-Nickel Steel Welds

The modern theory of intercrystallitic corrosion of stainless steels considers, as we saw, only two structural components of steel: austenite and carbides. However, another phase, namely primary δ-ferrite, can be present in welds. It has been said above that in oder to eliminate hot cracks in welding of austenitic steels it is absolutely necessary to have this structural compoment in the weld. What is its role in the processes of intercrystallitic corrosion?

Payson, who back in 1931 /193/ was one of the first to call attention to the beneficial role of ferrite, assumes that intercrystallitic corrosion in two-phase steels can be prevented owing to the fact that ferrite seems to be enriched with carbon and that it is precisely here, inside the ferrite sections, that chromium carbides are formed. As is known, there is less carbon in ferrite than in austenite, and Payson's hypothesis concerning the concentration of carbides in ferrite cannot be considered as founded.

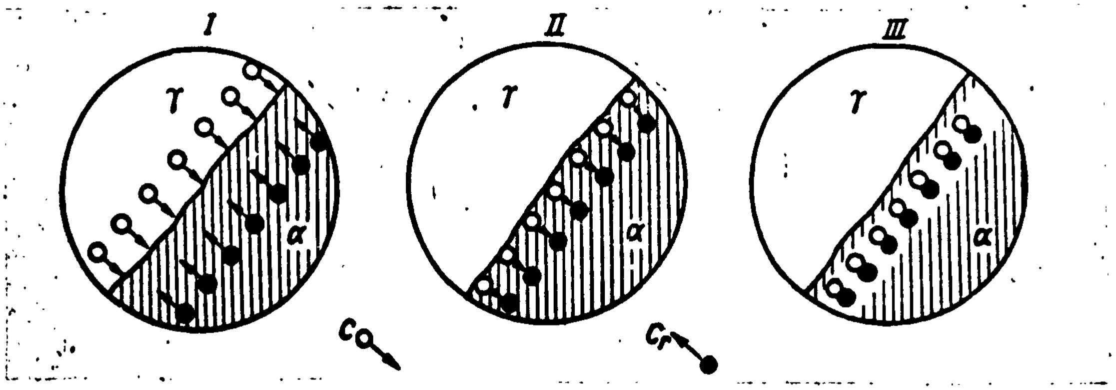

Fig. 92. Diffusion processes taking place on the phase boundary between austenite and primary ferrite in a weld (schematic); I, II, III — stages in the process of two-sided diffusion of C and Cr.

Investigations made at the Institute of Electric Welding permit the following answer to the question of the beneficial role of ferrite in welds.

On the basis of modern studies in the field of diffusion in solids it may be affirmed that diffusion processes in ferrite take place at a considerably gaster rate than in austenite. Ferrite (α-solid solution) differs from austenite (γ -solid solution) by a less compact packing of the atoms in the crystal lattice. The mobility of the atoms is higher in the body-centered lattice of the α (δ)-phase than in the face-centered lattice of γ-iron. Thus for example, according to data of Ya. S. Umanskiy and others /134/, the diffusion coefficient of nitrogen is 2000 times higher in ferrite than in austenite. This study, as well as studies by S. D. Gertsriken, I. Ya. Dekhtyar and others, show that not a single case is known where the diffusion rate is higher in γ-iron than in α-iron. The latest investigation made by P. L. Gruzin and others /26/ with the aid of the radioactive isotope C^{14} established that at 850°C carbon diffusion is 15 times faster in ferrite than in austenite. It may be affirmed that chromium also diffuses faster in ferrite than in austenite.

Let us examine the phenomena taking place on the phase boundary between austenite and primary ferrite in a weld under the effect of critical temperatures (Fig. 92). Primary ferrite, owing to dendritic liquation, differs from the austenite that surrounds it by a higher chromium concentration. During slow cooling after crystallization of the weld or under the repeated effect of critical temperatures the carbon atoms travel from the austenite into the ferrite. This is promoted by three factors: 1) higher carbon concentration in austenite than in ferrite; 2) accelerated carbon diffusion during the transition through the phase boundary between austenite and ferrite, owing to the higher diffusion rate in ferrite; 3) the tendency of carbon to combine with the active carbide-forming element chromium, in accordance with the laws of reactive diffusion (there is more chromium in ferrite than in austenite).

It must be assumed that the process of reactive diffusion plays the dominant part, since the diffusion rate of the carbon atoms, which, together with the iron base of the weld form an interstitial solid solution, considerably exceeds the diffusion rate of the chromium atoms which together with the iron form a substitutional solid solution.

The peculiar structure of clad steel (MSt.3 + 08Kh12) discovered by us together with Ye. S. Kudelya and A. S. Dem'yanchuk /70/ indicates that diffusion of carbon atoms from a metal not containing chromium into a metal

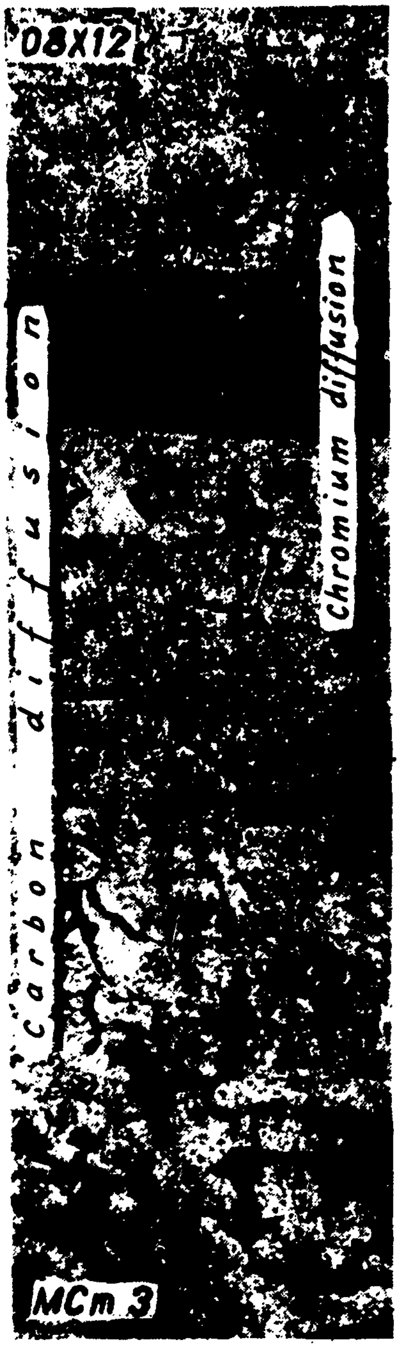

Fig. 93. Carbon diffusion discovered in clad steel with a ferritic chromium cladding.

that does contain chromium is entirely possible. Clad steel is produced by the joint hot rolling of two steel ingots, one of ferritic carbon, the other of ferritic chromium steel. The carbon steel contains not more than 0.15% carbon and up to 0.2% chromium, whereas the chromium steel contains 12% chromium and up to 0.07% carbon. During rolling at a temperature above the A_{c_3} point, the low-carbon steel acquires an austenitic structure, while the low-carbon chromium steel maintains its ferritic structure. Owing to the difference in the chromium and carbon concentrations in the steels and to their different crystal lattice, the carbon atoms, in compliance with the laws of reactive and concentrational diffusion, travel from the carbon steel not containing chromium into the ferritic chromium steel. As a result, the layer of MSt.3 steel which is in contact with the chromium cladding looses almost all of its carbon to a depth of 1 mm, whereas the layer of the chromium steel becomes, on the contrary, extremely rich in carbon; the carbon concentration in the chromium steel immediately adjacent to the MSt.3 steel increases almost 15 times and reaches 1% (Fig. 93). Together with carbon diffusion, an opposite process of chromium diffusion takes place from the chromium cladding into the MSt.3 steel. Although chromium diffusion is difficult in austenite, as compared to ferrite, the chromium atoms nevertheless travel through the boundary between the two steels owing to the great difference in the chromium concentration.

Reactive carbon diffusion into a metal with an elevated content of carbide-forming elements is also possible if this metal has a higher carbon content than the metal not containing any carbide-forming elements. In other words, ascending carbon diffusion is possible when metal of such different composition are in contact. Therefore, carbon diffusion takes place also in such clad plates in which the St.3 steel and the 08Kh12 or 18-8-type steel cladding have the same carbon concentration /17/. It is also known that, in the case of prolonged heating of weld joint of unalloyed steel with austenitic chromium-nickel welds, carbon, owing to its tendency to combine with chromium, diffuses from the zone adjacent to the weld into the weld metal.

Together with the process of displacement of carbon from austenite into primary ferrite in the weld under the effect of critical temperatures, there is apparently also a displacement of chromium atoms from the inner zone of each ferrite section towards its periphery. The carbon atoms are continuously encountered by new chromium atoms.

Thus, the diffusion processes in the weld result in changes similar to the changes taking place in rolling and heating of clad steel. Carbon atoms coming from austenite and chromium atoms coming from the central parts of the ferrite sections accumulate near the boundary surface between austenite and ferrite, i. e., inside and on the boundaries of the ferrite sections. Consequently, the formation of chromium carbides or some other excess phase including such carbides takes place inside the ferrite sections and on their boundaries. We assume that in austenitic-ferritic welds in which, as we know, the intercrystalline films are extremely thin, precipitation of the excess phase on the boundaries of the austenite grains (crystals) is extremely low. The processes and phase transformations taking place at the points where primary ferrite is situated become of primary importance.

It follows from what has been said above that the presence of primary ferrite in welds introduces qualitative changes into the processes, which can cause susceptibility of the metal to intergranular corrosion. The main feature of these qualitative alterations is the localization of phase transformations in those sections of the weld that are occupied by primary ferrite.

It is natural to assume that this localization must have a beneficial effect on the corrosion resistance of the weld, and this effect will be the stronger, the finer and more declicate the microstructure of the weld.

In a purely austenitic single-phase weld (Fig. 94, a) with an undeveloped crystal surface and with thick intercrystalline films precipitation of the excess phase takes place in the form of continuous chains along the boundaries of the acicular austenite crystals. This results in the formation of continuous tracks of austenite highly empoverished of chromium. The conditions for penetration of the aggressive reagent into the weld are thus made favorable. Under these conditions, intercrystallitic corrosion develops very strongly (Fig. 94, a).

In two-phase austenitic-ferritic welds (Fig. 94, b) with a developed crystal surface and with a refined disoriented structure, primary ferrite is deposited in the form of individual ramified petals. The bulk of primary ferrite is formed during crystallization inside the dendrites, and only a small part on their boundaries. Therefore, the major part of ferrite in the frozen weld is situated inside the columnar crystals, and not on the grain (crystal) boundaries. As a result, precipitation of the excess phase rich in chromium takes place not only on the grain boundaries, but also inside the

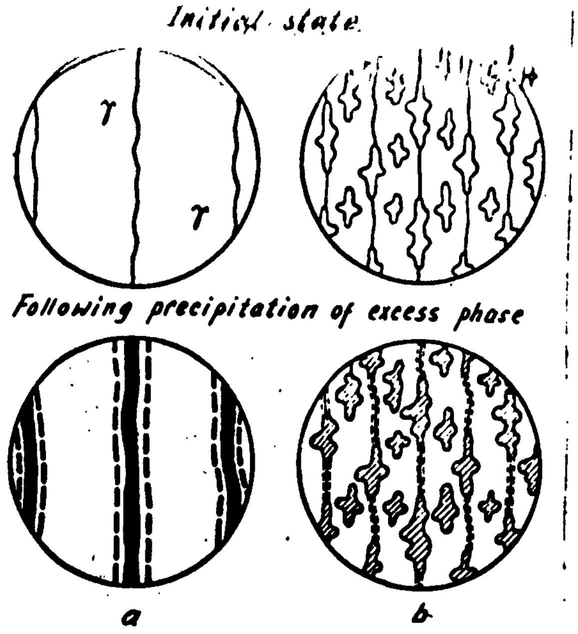

Fig. 94. Precipitation of the excess phase in the weld:

a) austenitic single-phase weld; b) two-phase austenitic-ferritic weld.

grains. Impoverishment of chromium is localized in the sections where the ferrite component is situated. Thus, the chemically unstable sections which border the ferrite petals alternate with austenite sections in which no phase transformations took place. The aggressive reagent will encounter resistant obstacles on its way toward the depth of the weld. As a result, the development of intergranular corrosion is, to a certain extent, made difficult in austenitic-ferritic welds.

The beneficial role of primary ferrite in stainless steel welds does not end in the localization of chromium impoverishment of austenite. The presence of ferrite sharply increases the capacity for immunization of welds, in comparison with rolled or forged steel of the same composition.

Experiments made for determining the immunization period T_i showed that austenitic-ferritic 18-8-type steel welds containing 0.09% carbon without any stabilizers such as titanium or niobium become absolutely immune to intergranular corrosion already after 18 to 24 hours of soaking at 650°. It turns out that T_i for a weld at 650° is lower by a factor of several tens than in the case of rolled steel of similar composition. The rapid immunization of welds is undoubtedly the result of the sharp increase in the diffusion rate of carbon and chromium in the weld, as compared with steel. The diffusion processes in the weld metal are accelerated by three circumstances: 1) the distortion of the crystal lattice, caused by internal stresses arising in the weld as a result of the rapid heating and cooling of the welded part and rapid primary crystallization; these stresses attain the yield point; 2) the presence of a ferritic component; 3) a structure that is finer than in the steel.

While examining the effect of heat treatment on the weld structure (see page 88), we pointed out that the stressed state of the weld facilitates diffusion. This is evident from the relationship between the diffusion coefficient and the activation energy

$$D = Ae^{-\frac{Q}{RT}}, \qquad (18)$$

where D is the diffusion coefficient at an absolute temperature T;

R is the gas constant (1.987 cal/g atom);

A is a factor determined by the type of crystal lattice;

Q is the activation energy or diffusion heat necessary to free a metal atom from the lattice and make it mobile.

The diffusion processes are markedly accelerated and the immunization period T_i is shortened if the temperature is increased and also if the weld metal is simultaneously subjected to heating and tensile stress. The effect of plastic deformation hardly yields to that of cold plastic deformation, as far as efficiency is concerned (see Section 7, Chapter V). Experiments showed that if a tensile stress of 20 kg/mm^2 is created in an austenitic-ferritic weld, i. e., a stress that is below the yield point, and if the weld is at the same time heated to 850°, even if only for a short time, for example by a moving gas torch, the immunization period can be reduced by several times. These experiments make it possible to affirm that service stresses and high service temperatures can promote rapid immunization of welds in real structures.

The influence of the ferrite component on the diffusion rate of carbon and chromium depends on the character of ferrite distribution in the austenitic weld base. It is natural that an increase in the contact surface between austenite and ferrite must speed up diffusion and reduce the immunization period. Therefore, austenitic-ferritic welds with a disoriented refined structure have the shortest immunization period. Thus, for example, the immunization period T_i of austenitic-ferritic double welds containing 0.10 to 0.12% C and not containing titanium and niobium, amounts to about 50 hours at 650°, 4 hours at 700° and less than 2 hours at 850°. The following values of T_i were obtained for austenitic welds: about 200 hours at 650° (more than 4 times more) and about 50 hours at 700°C (more than 12 times more)/63/. According to data of /254/ even 5000 hours of heating at 550° do not cause immunization in 18-8-Nb-type welds, but immunization is achieved already after 500 hours at 650°C.

Welds with a small cross section and a finer structure are immunized at a faster rate than single-pass welds with a large cross section, the chemical composition being the same and the quantity of δ-phase approximately equal.

In welds with a refined structure the decrease in the chromium content in austenite is less than in welds with an oriented structure or in welds containing a small quantity of ferrite. In order to eliminate the susceptibility of welds to intercrystallitic corrosion, it is necessary to increase the chromium content in the empoverished sections of austenite, to be precise at least up to the stability treshold, i. e., to 12.5 atomic %. The smaller the difference between the chromium content near the ferritic sections and the stability limit, the faster the achievement of secondary corrosion resistance (after immunization) of the weld.

The stated ideas show that primary ferrite has a beneficial effect on the corrosion resistance of welds owing to: 1) localized impoverishment of chromium, and 2) a sharp increase in the susceptibility to immunization.

4. Methods of increasing the corrosion resistance of welds of chromium-nickel austenitic steels.

The following measures of protecting intergranular corrosion of stainless steels have been developed in practice:

1. The greatest possible decrease in carbon concentration to the point of its solubility in austenite at normal temperature, i.e., to between 0.02 and 0.03%. Such steels are called ultralow-carbon stainless steels.

At this low-carbon content, there is no excess-phase precipitation, and, hence, the causes of susceptibility to intergranular corrosion are eliminated.*

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* According to literature data, American ultralow-carbon stainless 18-8 ELC and 18-8L steels lose their corrosion resistance after prolonged heating at 485 to 535°C. The corrosion rate in nitric acid reaches $51 \cdot 10^{-3}$ inches annually /167/, i.e., about 1.4 mm/year.

2. Hardening steel from a temperature between 1050 and 1150°C. Such heat treatment causes the excess phase to dissolve, and fixes the single phase austenitic structure of the steel. If heat treatment at critical temperatures is repeated, such steel regains its tendency to corrosion.

3. To include in the composition of a steel, elements of higher carbide-forming activity than chromium (so-called stabilization). These elements combine with carbon to form stable carbides. Hence, excess phase precipitation does not lead to any dangerous decrease in chromium content. As a result, steel remains unsusceptible to intergranular corrosion after exposure to critical temperatures.

Carbide-forming ability of an element may be judged by the magnitude of free energy liberated in the corresponding reactions.

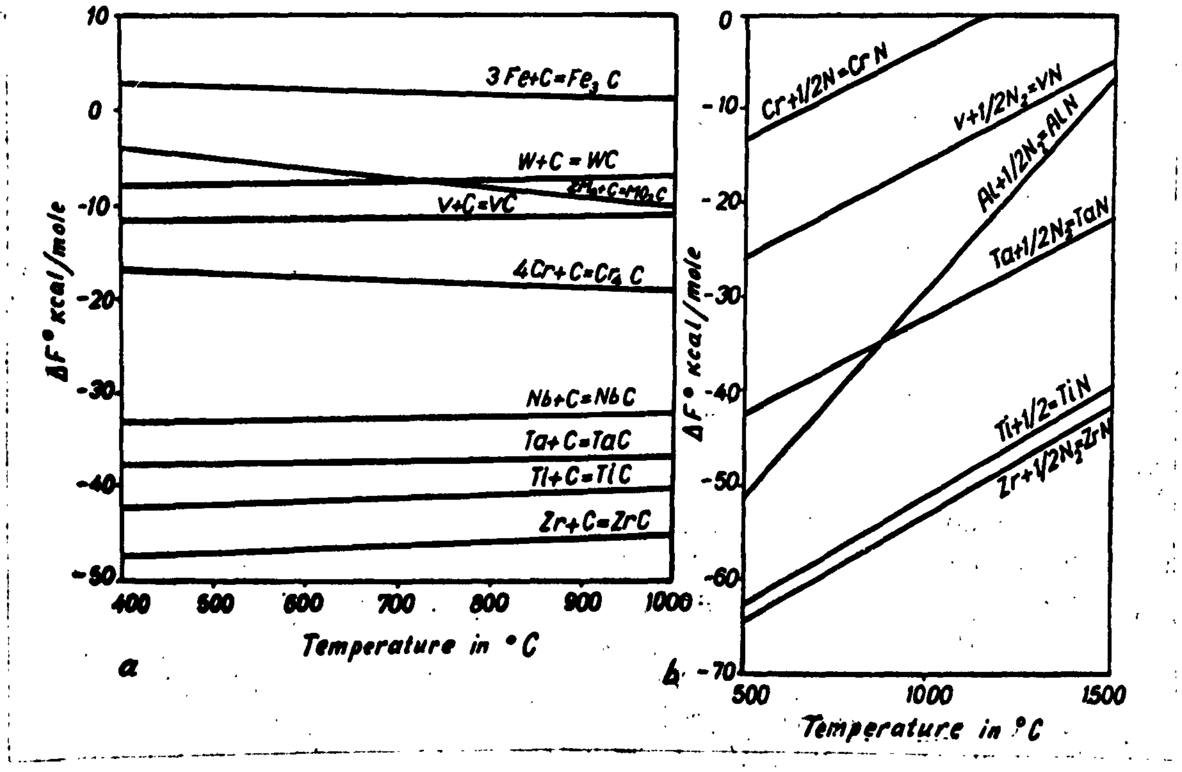

Fig. 95. Free energy, released in a) carbide formation, b) nitride formation, as a function of temperature.

Figure 95, a, presents data (calculated by us from the thermodynamic constants as cited in the work of O. Kubashevskiy and E. Evans) on the variation of free energy released in carbide-forming reactions of a series of elements, as a function of temperature. According to these data, zirconium, titanium and niobium are the most effective carbide-forming elements within the critical temperature range. Heretofore, only titanium and niobium plus tantalum are used for the stabilization of austenitic steels. The high affinity of zirconium to oxygen makes its admixture to steel difficult.

Modern techniques of vacuum fusion will probably result in zirconium being utilized for the stabilization of austenitic steels.

In addition to these three methods of protecting austenitic steels from structural corrosion, N. D. Tomashov /130/ recently proposed a procedure involving small additions of platinum and palladium. As components of austenitic steel, these elements create microcathodes in the latter, which ensures easy passivation of the steel. A lower anode activity of the alloy reduces the rate of corrosion. According to data /130/, 1Kh18N9 steel corrodes in 40% H_2SO_4 at a rate of 18 gr/m^2 • hr. The addition of a mere 0.1% platinum or palladium to the steel reduces its rate of corrosion 10 to 15 times. This effect is still more striking in the case of Kh27 steel: here, an addition of 0.5% Pd reduced the corrosion rate 300 times. In work /130/, silver is referred to along with Pt and Pd. However, our attempts to alloy welds of 1Kh18N9T steel with silver proved that due to its complete insolubility in iron and nickel, silver emerges to the weld surface in the shape of globules. Additions of silver in the form of an alloy adsorbable by the weld pool would probably serve the purpose better. To our knowledge, experiments on the alloying of welds with Pt and Pd have not yet been made.

Let us examine the use of 1Kh18N9T steel of normal carbon content (0.12% C) to see in how far the above-mentioned methods of preventing intergranular corrosion prove effective when applied to welds.

Suppose an Sv-0Kh18N9 ultralow-carbon wire of 0.03% carbon content is used for welding. For carbon-content determination, we will use Eq. (9). Since, in the welding process of austenitic steels carbon practically is not subject to oxidation ($\Delta C = 0$), this equation takes the form:

$$[C]_{\underline{w}} = \gamma [C]_{\underline{m}} + (1 - \gamma)[C]_{\underline{e}}^{\;*}$$

In manual welding, the basic metal does not exceed 20%, and, hence

$$[C]_{\underline{w}} = 0.2 \cdot 0.12 + 0.8 \cdot 0.03 = 0.048\%.$$

When carbon content is that low, a weld in natural condition is sufficiently corrosion-resistant. N. Yu. Pal'chuk has whown that the quantity $\underline{T}_1$ increases appreciably with a decreasing carbon content in manual welds of 18-8 steel, and that even after repeated and relatively long exposure to critical temperatures, such as in multipass welding, a tendency to intergranular corrosion need not be feared.

In automatic flux welding, γ is usually about 70%. Here,

$$/C/_{\underline{w}} = 0.7 \cdot 0.12 + 0.3 \cdot 0.03 = 0.09\%.$$

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* Translator's note: The subscripts used in this equation denote:

$\underline{w}$ = weld, $\underline{m}$ = metal, $\underline{e}$ = effective.

Experiments have shown that a one-sided single-pass automatic weld with this carbon content, in natural condition, is sufficiently resistant to intergranular corrosion, but that it swiftly loses this resistance under the effect of dangerous temperatures. Thus, for example, if a layer is welded to the opposite side of a double weld, the latter's resistance to structural corrosion may be lost.

Thus, we can see that, although there is some point in the application of ultralow carbon electrodes, it does not in itself present the solution of the problem how to make welds resistant to corrosion. This applies especially to automatic square butt welding.

In practice, quenching of welded constructions for austenite is in the vast majority of cases unfeasible because of the tremendous technological difficulties involved. Moreover, over-all heating to hardening temperature causes deformation of the structural shapes, while local heating results in concentrated stresses.

Stabilization of welds with titanium is possible in gas-shielded welding (argon, helium) and also fluorine-flux welding. In unshielded arc welding with thickly coated electrodes, titanium oxidizes so intensely as to greatly impede alloying of welds with this element. Welds alloyed with titanium in amounts large enough to resist intergranular corrosion, have a relatively low corrosion resistance in nitric acid (see below). This limits the use of titanium for alloying of welds in practice. Using niobium for stabilization of welds is very cumbersome due to its tendency to form hot cracks in welds. As was pointed out in Chapter IV, special techniques are required to neutralize the harmful effect of niobium.

Hence, none of the procedures applied to counteract intergranular corrosion in stainless steels may be utilized in welded constructions of these steels without taking additional preventive measures.

On the basis of investigations conducted at the Institute for Arc Welding, the following new methods were proposed for resisting intergranular corrosion in welds of austenitic chromium-nickel steels: 1) alloying the weld metal with ferrite-forming additions (titanium and niobium excluded) with a view to obtain an alloyed primary ferrite of high chromium content in the weld, and 2) immunization heat treatment, or so-called stabilization annealing, of ferritic-austenitic welds. Both methods are based on the favorable effect of the primary ferrite,* discussed in Section 3.

5. Effect of Alloying Admixtures on the Resistance to Intergranular Corrosion of Welds of Austenitic Steels.

The effect of alloying elements on the resistance to intergranular corrosion of welds directly depends on the nature of their action upon the primary structure of the weld. Experiments revealed that ferritizing elements enhance the corrosion resistance of welds, whereas austenitizing admixtures diminish it.

Ferrite-forming admixtures. Both theoretical and practical interest is focused primarily on studying the effect of aluminum and silicon, since, contrary to most ferritizers, these elements do not form carbides. On the example of welds alloyed with aluminum it is possible to obtain a clear view of the effect of the primary ferrite on the acceleration of diffusion processes and on the prevention of intergranular corrosion.

Experimental results show convincingly that the ferritizers (aluminum and silicon) are not inferior to the carbide-formers (titanium and niobium) in making welds resistant to intergranular corrosion /63/.

It is appropriate to mention here the entirely false opinion which until quite recently prevailed in the pertinent literature, claiming that the carbide-formers titanium and niobium are alone capable of preventing intergranular corrosion in welds.

In fact, until recently modern literature on welding of stainless steel entirely ignored diffusion in the weld, while the effect and significance of the primary structure and of the primary ferrite were not paid the attention they deserve.

This attitude of welding experts towards ferrite has deeply rooted, stemming from the automatic acceptance of the view held among metallurgists that the α (δ)-phase in austenitic steels was undesirable. Since the

α-phase impedes mechanical hot working of austenitic steels, the attitude of the metallurgists toward it was not unfounded; however, the welders' view lacked foundation, as developments have shown.

Experiments have proved that the beneficial effect of ferrite-forming elements - titanium and niobium included - on the resistance of welds to intergranular corrosion increases appreciably when oxygen-free and low-silicon fluxes are used. Apparently, this is to be explained by the modifying refining effect of these fluxes on the primary structure of welds.

The effect of aluminum in speeding up diffusion and shortening the immunization period T_{im} can be judged by the results of the following tests. Specimens with welds containing 0.11% carbon and 0.4% aluminum were tested for intergranular corrosion by boiling for 72 hours after differential holding at 650°C (Table 43).

Table 43

Occurrence of intergranular corrosion after differential holding at 650°C and 72 hour boiling in sulfuric solution of blue vitriol

Soaking period of welded specimens at 650°C in min.						
0	2	5	10	15	25	30
None	Present	Present	Present	Present	Present, none	None

Experiments showed that a double weld of 18-8 steel containing 0.11% carbon and no aluminum is immunized only after 50 to 60 hours of soaking at 650°C.

An addition of 0.4% aluminum to a weld of similar composition permits of shortening the immunization period more than hundredfold, reducing it to 25 - 30 min. The favorable effect of aluminum as, indeed, of all other ferritizers -- niobium and titanium included -- should not be interpreted as being merely due to the formation of the ferrite component as such. There can be no doubt that, similar to the resistance of welds to hot cracking, it is not so much the amount of the δ-phase but rather its properties -- more properly its qualitative characteristics -- that are of decisive importance. Of these, the ability of ferrites to accelerate to a greater or less extent the diffusion of chromium and carbon is, probably, most significant. It can be readily conceived that a speedier diffusion of chromium and carbon in the δ-phase must undoubtedly result in a more rapid immunization of the weld. When alloying welds with ferrite-forming elements, the extent to which the diffusion of chromium and carbon in austenite is quickened or slowed down must also be of some consequence(to be sure, not only the δ-phase, but also the austenite is alloyed with ferritizers).

V. I. Arkharov /7/ notes that aluminum appreciably increases the diffusion rate of chromium in austenite. According to G. N. Dubinin's data, molybdenum, tungsten, chromium and silicon have the same effect, while the diffusion rate of carbon in austenite is slowed down by nickel, manganese and copper. Taking these data into account, it can be understood why welds of 18-8 steel additionally alloyed with the ferritizers Al, Mo, W, Cr, and Si, are more swiftly immunized than those with a higher Ni, Mn, and Cu content, other conditions being equal. In the former case, secondary resistance will be achieved more rapidly due to the chromium-content being evenly distrubuted. In the latter case, this process is delayed, since the formation of carbides takes a longer time.

The favorable effect of the δ-phase is also apparent from the fact that smaller amounts of titanium and niobium are needed for stabilizing welds of 18-8 steel than in the case of rolled or forged steel. For the stabilisation of steel with titanium or niobium, these elements should be added in amounts not smaller than as calculated from Formulas (2) and (3). Practice shows that the stabilisation of a weld of ferritic-austenitic structure with a carbon content of 0.10% to 0.12% is achieved already by an addition of 0.5 to 0.6% niobium or 0.30 to 0.35% titanium, as against the 0.8 to 1.0% and 0.4 to 0.5%, respectively, required by these formulas.

Along with aluminum and silicon, other ferritizers also have the ability to stabilise the welds of stainless steels. The effect of vanadium, tungsten and molybdenum can be determined from the data obtained in 1950 which we present in Table 44.

It should be noted that in a paper published in 1955 Baerlecken and Hirsch /154/ who studied the effect of various ferritizers on the resistance of austenitic steels to intergranular corrosion, arrived at exactly the same conclusions about the faverable effect of vanadium and molybdenum when used in the said concentrations.

Among the ferritising elements, vanadium is the most effective next to zirconium, niobium and aluminium. Excellent results have been obtained in the alloying of welds by using silicon in combination with vanadium.

Table 44

Effect of vanadium, tungsten, and molybdenum on ~~the~~ resistance to intergranular corrosion of 18-8 steel welds

Stabilizing admixture		Chemical composition of weld metal in %					Presence of intergranular corrosion after 2-hr heating at 650°C and 72-hr boiling
Elements	%	C	Si	Mn	Cr	Ni	
Vanadium	0.6	0.10	0.67	0.68	18.2	9.7	Present
	1.26	0.10	0.87	0.80	19.3	10.8	None
Tungsten	0.88	0.10	0.54	0.82	17.7	9.1	Present
	1.48	0.11	0.62	0.80	18.2	9.4	None
Molybdenum	1.68*	0.11	0.54	0.69	18.5	8.1	Present
	2.30	0.10	0.62	0.70	19.7	9.4	None

* Intergranular corrosion is absent after 8-hr heating at given temperature.

In automatic flux-welding, a chromium-nickel-silicon-vanadium electrode, trade mark OKh18N9F2S (EI606), warrants a high resistance of welds to intergranular corrosion, which is by no means inferior to that obtained in alloying welds with niobium. The same results are obtained in manual welding with silicon-vanadium-coated electrodes /36/.

Vanadium is not only an active ferritizer, but forms carbides as well (Fig. 95, a). This may raise objections as to the correctness of our interpretation of the favorable effect of vanadium as a ferritizer on an increased resistance of welds to intergranular corrosion. It can be stated that the effect of vanadium does not necessarily imply the formation of a ferrite, being similar to that of other carbide-formers (titanium, and niobium)that is to say vanadium binds carbon, prevents formation of chromium carbides, and, by so doing, prevents susceptibility to corrosion. Indeed, in the works /27, 137, 172/ the effect of vanadium is evaluated in just this way.

Tests, specially conducted for this purpose, proved the correctness of our point of view. Welds were obtained with different contents of nickel (8 to 13%), the concentration of all other elements, including chromium and carbon, being equal. Each of the two welds had a vanadium content of 2%. The weld of 18-8 steel, whose ferritic-austenitic structure was due to vanadium, proved unsusceptible to granular corrosion, whereas the weld of 18-13 steel, maintained its austenitic structure regardless of the presence of 2% V, and proved to be unresistant. At the same time, purely austenitic welds that were alloyed with titanium and niobium (that is, with carbide-formers of greater intensity than V) in amounts sufficient for their stabilization (about 4 per cent Ti and about 0.7 per cent Nb at a carbon content of 1 per cent), are resistant to intergranular corrosion. It is noteworthy that in rolled steel, vanadium behaves in the same manner (Table 45) /205/.

Due to its low nickel content, specimen # 1 acquired a ferritic-austenitic structure and became corrosion-resistant after vanadium had been added.

Influence of vanadium on intergranular corrosion resistance of chromium-nickel austenite steel [205].

Nº of specimen	Chemical composition of steel, %				Microstructure	Susceptibility to intergranular corrosion after heati for an hour at 600°
	C	Cr	Ni	V		
1	0.12	18.09	8.94	2.40	Austenite + ferrite	None
2	0.10	17.37	12.41	2.32	Austenite	Extremely high
3	0.12	17.35	14.05	3.01	Austenite	" "

Note. Silicon and manganese contents in the steels are practically equal.

Thus it is found that ferrite-forming admixtures, even the most active ones such as aluminum, can ensure intercrystallite-resistant welds only when the two-phase austenite-ferrite structure is formed. Incidently, this accounts for the well-known fact of ~~the~~ susceptibility to intergranular corrosion of welds with high nickel content in steels of the 18-14 type containing up to 3 % ferritizer-molybdenum but preserving their austenite structure. The part of molybdenum is described in [259].

Since the decrease of chromium content in austenite is the cause of intercrystallite corrosion, efforts to stabilize the welds by increasing chromium content in them seem~~s~~ entirely natural. Such a suggestion may also, however, ~~arise~~ raise al-se objections. It is known that steels of the 25-20 type and their ~~seams~~ welds are considerably more susceptible to inter~~crystallite~~granular corrosion than steels of the 18-8 type and their ~~seams~~ welds, despite a nearly 1.5 times greater chromium content in the former*. Experiments showed that with the increase up to 22 % in chromium content, a ~~seam~~ weld with 8 % nickel becomes immune to inter~~crystallite~~granular corrosion, despite the absence of titanium and niobium in it [63]. How ~~could~~ can then the fact be explained that ~~seams~~ welds of the 25-20 type steel are less corrosion-resistant than those of the 25-20 type steel? This, evidently, is also linked with the nature of the primary microstructure of the ~~seam~~ weld. In a purely austenite ~~seam~~ weld of the 25-20 type steel, the separation of the excessive phase ~~occurs~~ in the form of whole chains and is accompanied by a great loss of chromium in austenite (Fig. 94a). In the seam with 8 % of nickel content the increase in chromium content resulted not only in the increased amount of ferrite but also in the increased chromium concentration in it. The rate of diffusion increased so greatly that the separation of the excessive phase no longer resulted in impoverishment of the austenite; in place of the chromium atoms spent to form the excessive phase, new ones arrive instantly at the austenite-ferrite boundary and the concentration of chromium does not drop below the stability threshold.

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* The authors of reference [227] claim that steel of the 25-20 type is seemingly less susceptible to inter~~crystallite~~granular corrosion than steel of the 18-8 type because of the more uniform distribution of chromium carbides.

Nearly three years after the publication of our data on the possibility to prevent intergranular corrosion in welds of the 18-8 steel by increasing chromium content up to 22% without resorting to niobium [63], an analogous work was carried out in the USA. In December 1953, new electrodes, made from niobium-free 22-10 steel, were proposed there for the manual welding of 18-8 steel [159]. The welds that were obtained with the use of these electodes contained 22% chromium; they were of two-phase structure (10 to 15% δ-phase) and resisted common and intergranular corrosion.

We investigated the influence of such ferritizers as beryllium, tin, antimony and phosphorus. Antimony and phosphorus cause an intensive cracking of welds. Beryllium, when its content in the weld is about 0.50%, prevents intergranular corrosion but impairs the plasticity of the metal of the weld.

Tin, introduced in amounts up to 2% into a weld of 18-8 steel, fully prevents intergranular corrosion. Small additions of lead increase the positive effect of tin. Additions of tin (greater than 2%), however, cause hot cracks, due to the formation of easily melted intercrystal layers. Still, due to the rapid increase in the amount of the Ni - Sn eutectic, a further increase in tin content prevents cracks. Experiments with beryllium and tin additions are fundamentally important, since they confirm the correctness of our views as to positive part of alloyed ferrite.

According to Wiles, sulfur can prevent intergranular corrosion in welds of 18-8 steel during torch welding, if hydrogen sulfide is added to the acetylene. Welds obtained in a usual way show intergranular corrosion. Intergranular corrosion no longer occurs in the weld, if 5% sulfur is intruduced in the welding torch. Analogous results were obtained in welding with sulfur-containing coatings. Applying sulfur in flux welding we succeeded in avoiding intergranular corrosion; in this case, however, the overall-corrosion resistance of the weld is markedly impaired in oxidising liquid madia (see below). Antimony acts analogously. Welds should not be alloyed with these elements in practice.

Austenite-forming admixtures. Nickel, manganese, copper and cobalt impair the intergranular -corrosion resistance of welds, causing austenization of their structure. It should be noted that in the presence of titanium and niobium in the weld, and also zirconium in amounts sufficient to neutralize carbon [see Formulas (2) and (3)], purely austenitic welds may be intergranular-corrosion resistant as a matter of principle. As a rule, however, such welds are affected by hot cracks and may not be permitted in welded heavy-duty structures. Speaking of the negative effect of austenitizers, we mean welds, free of titanium, niobium and zirconium carbide-formers. The mechanism of the negative action of austenitizers may be explained in this case as follows. The greater the amount of austenite and the smaller the amount of ferrite in the weld, the more unfavorable are the conditions in which are found chromium-impoverished regions. The cathode-austenite base in such welds has a considerably larger surface than in austenite-ferrite welds, which stimulates the corrosion current. The disappearance of ferrite reduces the diffusion rate. The negative effect of nickel, manganese and copper may be judged from their effect on the immunization period of welds. T_{im} increases considerably as the concentration of the said elements increases. So, for example, an increase up to 6.5% in manganese content or 3% copper additions increase T_{im} from 40-50 up to 200 hrs or more at 650°. The negative effect of Mn, Ni and Cu is also linked with their ability to slow down carbon diffusion in austenite.

The negative effect of carbon on the intergranular-corrosion resistance of welds is generally known. With the increase in carbon content the immunity of welds declines steeply. Thus, for example, an increase from 0.03-0.04% up to 0.10-0.12% in carbon content in automatic welds causes a nearly 20-fold increase in T_{im} at 650°.

The effect of carbon on the intergranular-corrosion resistance of manual welds in the 18-8 type steel may be judged from Tremlett's data [220] (Table 46).

In view of the harmful effect of carbon, electrode wires, even with traces of a graphite lubricant on their surface, may not be used. S. V. Junger has found that in welding with such wires, the weld, even in its natural state, does not resist intergranular corrosion. Welds, obtained with the same wire after removing graphite from their surface, were found well-resistant in their original state.

Table 46

Effect on intergranular-corrosion resistance of carbon content, duration of treatment at 650° and duration of boiling in a $H_2SO_4 + CuSO_4$ solution [220]

Carbon content in the weld, %	Occurrence of intergranular corrosion after tempering at 650°C for the following duration in minutes				
	2	3	4	5	6
	and after boiling for the following number of hours				
0.04	504, none	504, none	504, none	504, none	504, none
0.05	504, none	504, none	504, none	432, occurred	72, occurred
0.09	144, occurred	72, occurred	72, occurred	72, occurred	72, brittle fracturing
0.135	72, brittle fracturing along the weld when bent	72, brittle fracturing	72, brittle fracturing	72, brittle fracturing	the same

In welding stainless steels, each hundredth of a per cent of carbon already affects corrosion resistance of welds. Therefore any possible effort should always be made to reduce carbon content in the weld. The application of ferroalloys with higher carbon contents should therefore not be allowed in the manufacture of electrodes for manual welding.

With carbon and titanium total contents equal, the intergranular-corrosion resistance of the 18-8 steel may vary. There will be no corrosion, if all the carbon is combined in titanium carbides. However, if titanium is combined in part in nitrides or is in a γ-solid solution, steel may be susceptible to intergranular corrosion inspite of the fact that the ratio of the total titanium and carbon contents is in conformity with the stoichiometric ratio of these elements in carbide. Consequently, to prevent intergranular corrosion in 18-8 steel, it is important that the condition $\frac{Ti_{comb}}{C} \geqslant 5$ rather than $\frac{Ti_{tot}}{C} \geqslant 5$ be ensured, where Ti_{comb} is the titanium total content combined as carbides and Ti_{tot} is the titanium total content in the steel.*

As yet, there is no generally accepted view regarding the effect of nitrogen on the intergranular corrosion of welds. According to Binder, Brown and Franks, only 0.04% of nitrogen content impairs the intergranular-corrosion resistance of the 18-8 steel. At lower or higher concentrations, no harmful effect of nitrogen is observed.

One links the negative effect of nitrogen to its ability to form stable nitrides with titanium. In the presence of nitrogen in the weld, titanium in part is

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* V. F. Mal'tsev, V. Ya. Sych, "Steel" No 5, 1954.

spent to form nitrides and ~~its~~ the remainder is not sufficient to combine with carbon. Judging by the value of the free energy of formation of nitrides (Fig. 95b), the occurence of Zr, Ti, V, Nb, Al and Cr-nitrides in welds of the 18-8 type steel is most probable.

Other investigator[3] believe that the negative effect of nitrogen consists in its ability to form chromium nitride, CrN. They point out that at 600° the free energy of formation of chromium nitride has ~~the~~ a negative sign and its values in the formation of both chromium nitride and chromium carbide are of the same order. Proceeding from the fact of proximity of the values for the free energy of formation of titanium nitrides and its carbides, a conclusion is also drawn that titanium acts positively in two ways -- it combines with both carbon and nitrogen.

In our opinion, the effect of nitrogen on the corrosion susceptibility of a weld should not be regarded apart from its action on the structure. Nitrogen, whenever (in low concentrations) making the weld structure finer, must improve the weld's corrosion resistance.

Should the action of nitrogen result in of the weld austenization, the latter's resistance is bound to decrease. Experiments showed that nitrogen, indeed, has some effect on the corrosion susceptibility of a weld[10]. The action on nitrogen is positive in welding under pumiceous flux. It is known that a pumiceous flux in distinction from a glassy flux, provides a less perfect protection from air of the welding area. There is always more nitrogen in welds obtained under a pumiceous flux than in those made under a glassy flux. The intergranular corrosion resistance of a pumiceous flux is somewhat higher. In all, however, the corrosion effect of nitrogen both in welding under flux and with thickly coated electrodes is of inconsiderable practical significance.

Apparently, the effect of oxygen on the corrosion susceptibility of a weld is linked with its action on the structure (see page 72).

According to literature data, hydrogen does not increase ~~increases inconsiderably~~ intergranular-corrosion susceptibility to a significant degree.

Carbide-forming impurities. The positive effect of the most effective carbide-formers, titanium and niobium, is generally known. Among other impurities, more intensive than chromium as carbide-formers, zirconium should be mentioned. Our experiments with flux welding and N. Yu. Pal'chuk's experiments with manual welding showed clearly that this element prevents intergranular corrosion in welds of the 18-8 type steel. Its short supply and a great affinity ~~to~~ for oxygen still prevent zirconium from being widely used in practice as a stabilizer of welds. However, in electric-torch welding and non-oxygen flux welding, a rather complete zirconium recovery in the weld pool can be achieved. Our experiments showed that with the aid of zirconium intergranular corrosion can be prevented in welds in

Kh23N23M3M3, Kh23N28M3D3 and other stable austenitic steels.

6. Effect of Welding and Heat Treatment Conditions on Intergranular Corrosion Resistance of Welds in Austenitic Steels

Effect of welding conditions. In chapter II we noted that, depending on the conditions of welding, the primary structure of welds in austenitic steels undergoes substantial changes. The variations in conditions of welding which result in a finer primary structure of the weld further the latter's corrosion resistance. Experiments show that welds with a finer structure possess a higher intergranular-corrosion resistance than those of the same chemical composition but with a coarser structure. Thus, for example, a single-pass weld on a 10 mm thick 1Kh18N9T steel with about 1.0% content of vanadium resists intergranular corrosion not only in its original state but also after a two-hour treatment at a dangerous temperature (650°C). At the same time, the single-pass weld of analogous composition but obtained on a 50 mm thick steel and possessing therefore a considerably coarser structure (Fig. 29) is susceptible to intergranular corrosion even in its original state. In this case, the decline of the weld's corrosion resistance was also furthered, along with coarsening of its structure, by slower cooling, i.e., by prolonged action on the weld of offensive (critical) temperatures.

Slowed down cooling of welds in stainless steels results in the precipitation of the excess phase, which is absolutely inadmissible, if the welds, in their original state, should resist intergranular corrosion and will not be subjected to critical temperatures later on. Therefore, it is always urged in the literature on welding that the conditions and regimes for welding of stainless steals must be selected such that the cooling of welds proceeds at a maximum possible rate.

There exist different ways to accelerate the cooling of a weld in the process of welding. Acceleration is achieved with regimes distinguished by a low energy per unit length and a maximum possible rate of welding as well as by welding layers with small cross sections. It is recommended for the same purpose to weld stainless steels on copper backings cooled by ~~a~~ flowing water. Some authors propose that cold water be streamed on the weld root directly in the welding process, i.e., in the close vicinity of the arc /195/.

From thermal calculations, based on the well known N. N. Rykalin~~'s~~ works of /110/, it is possible to approximate ~~calculate in approximation~~ welding regimes which provide the required cooling rate of welds that undergo a repeated action of the welding cycle. In thermal calculations for structural-steel welding, C-shaped curves are applied, by superimposing them on the curves for the cooling in the vicinity of the weld. For stainless-steel welding, a curve for the dependence of the weld's corrosion resistance on temperature and time may be used in place of the C-shaped curve (see Fig. 91, a). A weld must be cooled at such a rate that the time during which the weld metal remains within the critical temperature range in the welding process is less than T_1 (Fig. 91, a). In this case, there is not enough time for the weld to become susceptible to intergranular corrosion /98, 99/.

N. Yu. Pal'chuk found that a temperature of 730°C is most dangerous for manual welds of the 18-8 steel. He found that at the most dangerous temperature, 730°C, the critical time T_1 (see Fig. 91, a) for this weld type, free of titanium and niobium stabilizers, practically does not depend on the manual welding regimes characterized, within the range from 2000 to 10,000 cal/cm^2, by the specific heat energy, and is entirely determined by the carbon content. With the decrease in carbon content from 0.15 to 0.09, T_1 increases exponentially from 10 to 12 sec for 0.15% C to 60 sec for 0.09% C. Given the value of T_1 for a weld with a given carbon content, it is possible to calculate such welding regimes at which, during a multi-pass manual welding, the total time of exposure of a working (first) layer of the weld-on metal is smaller than T_1.

Thus, in stainless steel welding the rate of cooling of the weld must be maximum. However, this rule is irrelevant, if the welds undergo immunizing heat treatment (after welding. In this case, ~~an~~ accelerated cooling of welds is undesirable and the welding process should be carried out in such a way that the crystallization and cooling of the welds proceed at the lowest possible rate. As a result, the most complete precipitation of the excess phase will occur in the weld already in the welding process.

Under the further action of critical temperatures, a single process of the equalization of chromium concentration will prevail in place of two - the formation of the excess phase accompanied by the impoverishment in austenite, and the diffusion of chromium into the chromium-poor sections of austenite. Hence, the period of immunization of slowly cooled welds is considerably shorter than that of rapidly cooled ones. This is also favored by the fact that when the crystallization is slowed down the structure of the weld becomes desoriented, while a directed structure is formed in a rapidly cooled weld. This may be illustrated as follows. Some 1Kh18N9T-steel specimens were single-welded ~~in~~ under conditions of extremely rapid cooling. For this purpose steel plates, previously cooled down to -70°C, were welded on a copper backing and the slag crust was removed immediately after the completion of welding. Other specimens were welded using the same addition agents, ~~at~~ under the same regime (current, arc voltage, welding rate), but ~~in~~ under conditions of slowed down cooling.

For this purpose, steel plates (preheated to 400°C) were welded on a flux backing, and the slag crust was removed only after the welded specimen had completely cooled. The specimens of both series were immunized, i. e., were kept for 4, 6, 12 and 24 hours at 700°C, and tested for inter~~crystallite~~ granular corrosion. It was found that slowly cooled welds turned corrosion-resistant already after a 12-hour treatment at 700°, while rapidly cooled ones remained corrosion-susceptible even after a 24-hour immunization (Fig. 96).

Thus, the evaluation of the effect of the rate of cooling on the corrosion resistance of stainless welds should depend on the actual conditions and should take into account the demands which must be met by the welds, and their subsequent heat treatment, i. e., whether or not the latter will take place.

It was pointed out in Chapter III that changes in the weld shape and in the direction of the heat outflow affect the location of the liquation zone. Apparently, the liquation zone is carbon-enriched and consequently it is this zone where inter-granular ~~crystallite~~ corrosion occurs most frequently. In narrow welds, distinguished by the presence of a plane in which crystals meet, the liquation zone is situated as shown in Fig. 97a. ~~In~~ Welds of such shape, usually occurring in single ~~one-side~~ welding, show a very intensive concentration of inter~~crystallite~~ granular corrosion along the longitudinal axis. Corrosion deterioration of the weld resembles in this case a sharp cut (Fig.

97 b). In a wide weld, the carbon-rich liquation zone is located in the upper axial part. Here the amount of ferrite is small, the structure is found to be directed and intergranular corrosion may develop /247/.

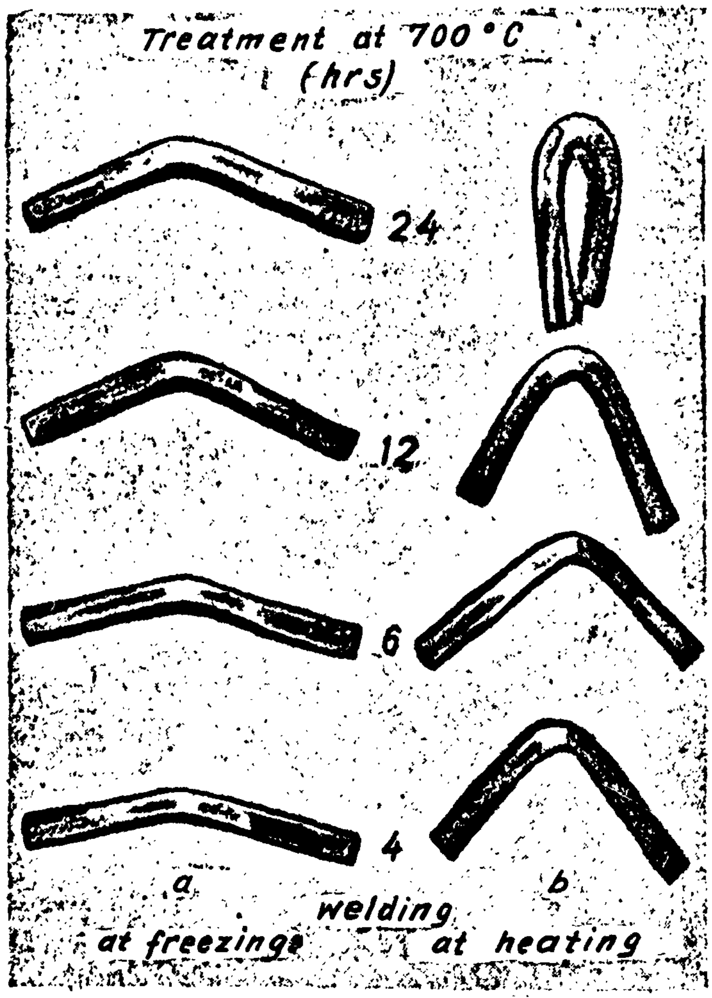

Fig. 96. Effect of cooling conditions during welding on the immunization rate of welded specimens at 700°C; a -- specimens welded under conditions of rapid cooling; b -- under conditions of slow cooling.

Effect of heat treatment. It has been noted above that the immunization period of welds is decreased considerably if tensile stresses act on the weld. We carried out comparative tests for intergranular corrosion on single-welded specimens. One portion of the specimens was given the usual oven treatment at 650 to 750°C while the other portion was put in a creep-test machine for treatment at the same temperature, but under stress. The ends of bent specimens were threaded for this purpose. These specimens were loaded in such a way that the tensile stresses on the weld metal were about 12 kg/mm^2. After exposure the welded specimens were boiled for 72 hours in a sulfuric acid solution of copper sulfate. It was found that the single welds, which underwent the combined action of tensile stresses (elastic deformation) and temperature, became immune ~~turned insusceptible~~ to intergranular corrosion in 4 hours at 700° and in 28 hours at 650°. The specimens which underwent the action of the above temperature without stress were found susceptible to intergranular corrosion. They became immunized as late as 24 hours after the treatment at 700°. Thus the presence of tensile stresses resulted in a nearly six-fold increase in diffusion rates ~~caused a nearly six times faster diffusion~~

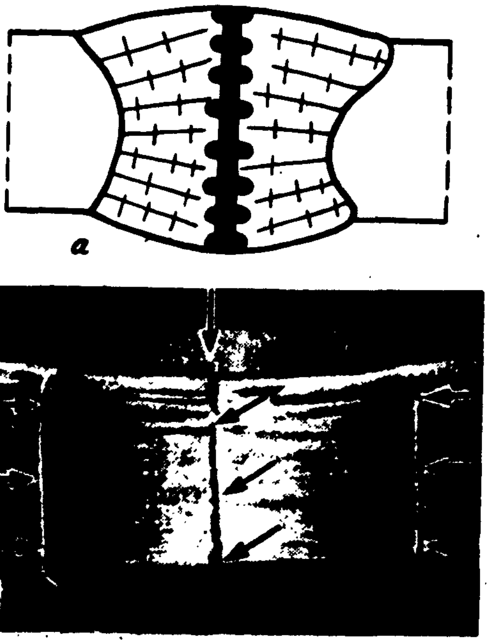

Fig. 97. Liquation zone (a), localized intergranular corrosion in the form of a sharp cut (b) along the longitudinal axis of the weld with a slack zone, and knife-line corrosion along the fusion line.

Metallographic and X-ray structural tests showed that the combined action of tensile stresses and the critical temperature results in an intensive decomposition of austenite and the formation of secondary ferrite. In our test, the weld in the as-welded condition contained 5 to 10% primary ferrite (δ-phase). Upon a 2-hour soaking at 700°C and the action of tensile stresses (11.1 kg/mm^2), about 50% secondary ferrite (α -phase) was found in the weld, while after a 4-hour heating it climbed as high as to about 70%. This is apparently what accounts for the intensified diffusion of chromium and reduction of T_{im}.

Quenching of welds from high temperatures is known to eliminate the susceptibility to corrosion, yet it has no practical application in the production of welded structures. The application of local heat treatment of welds, i. e., local quenching, if fundamentally realizable but still rare in practice. For local heating of welds of finished objects to the quenching temperature, high-frequency currents or acetylene-oxygen torch may be used. The application of local austenitising (hardening) may cause considerable stresses in the structure.

The above-mentioned peculiar properties of the austenitic-ferritic structure of a cast weld metal make it possible to effectively replace quenching from high temperatures by quenching from lower temperatures. A lower temperature of quenching is highly desirable, since it simplifies the heat treatment of welded structures, eliminates the danger of buckling and deformation, reduces scaling and relieves internal stress.

The following tests were carried out. Single-welded specimens were quenched in water at 650, 700, 850, 950, 1050 and 1150°C and then heated for 2 hours at a dangerous temperature (650°). The results of the subsequent bending tests, after a 72-hour boiling in a sulfuric-acid solution of copper sulfate are given in Table 47. The time of heating to quenching temperature varied from 30 minutes to 2 hours.

TABLE 47. Occurrence of intergranular corrosion after quenching, tempering at 650°, and boiling of welded specimens in sulfuric-acid solution of copper sulfate for 72 hours

	Quenching temperature in °C				
650	700	850	950	1050	1150
Occurred	Occurred	None	None, occurred	Occurred	Occurred

The data obtained reveal a positive effect of quenching from reduced temperatures on the corrosion resistance of welds. They show that when it concerns welds, quenching temperature may be reduced to 800-900°.

Basically, quenching from reduced temperatures is superior to usual austenite quenching in that the weld after such thermal processing ~~turns insusceptible~~ becomes immune to intergranular corrosion even ~~upon a repeated action of~~ under repeated exposure to dangerous temperatures.

We find that upon tempering at 650° the failure caused by intergranular corrosion in welds quenched from 1150°C is in no way less than in welds not quenched before tempering. Moreover, it may be supposed that welds quenched at 1150° may have become more susceptible to dangerous temperatures than those tempered at 650° without previous quenching. During heating at 1150°, there occur in the welds an equalization of the concentration of ferrite-forming impurities and a $\delta \rightarrow \gamma$ transformation, due to the diffusion taking place there. Water quenching secures the single-phase austenitic structure. There is no $\delta \rightarrow \gamma$ transformation during heating of welds at 850°, but only a moderate ferrite coagulation, along with the most complete precipitation of the excess phase; the chromium lost by the austenite is replenished through the diffusion of chromium from ferrite and from the interior of austenite grains. Water quenching secures in this case the two-phase austenitic-ferritic structure. Under the ensuing action of critical temperatures, the excess phase of a weld quenched from 1150° will precipitate following the pattern shown in Fig. 94a. There is usually no more precipitation of the excess phase after tempering at 650° in a weld pre-hardened at low temperature (850-900°). In an austenitic-ferritic weld which has not hardened at all, a tempering at 650° will result in the separation of the excess phase following the pattern in Fig. 94, b. As earlier noted, such welds will be more corrosion-resistant.

Special tests with automatic welds confirmed the accuracy of the suppositions set forth above (Fig. 98).

N. Yu. Pal'chuk proved that stainless manual welds, quenched at high temperature, become more susceptible to intergranular corrosion at dangerous temperatures than those not quenched before tempering. He suggests the following explanation of the negative effect of heating after quenching. During the process of crystal-

Fig. 98. The effect of hardening on the resistance to intergranular corrosion in instances of subsequent quenching:

a) weld, subjected to hardening; b) weld, not subjected to hardening following welding.

lization of a weld there occurs microscopic carbon liquation. Complex carbides are formed in separate carbon-enriched regions during the cooling of the weld. Consequently, the amount of carbon, which is capable of forming carbides under the ensuing action of critical temperatures, decreases. All the carbon is converted into a solid solution by quenching. At dangerous temperatures, the precipitation of carbides is furthered on account of the increasing concentration of carbon in the solid solution. As a result, the critical time of the hardened weld (T_1 in Fig. 91, a) falls below T_1 of the weld in the natural state /98/.

Experiments showed that rapid cooling of welds in water is unnecessary after heating at 850°. Positive effect is reached with cooling in air as well. Such heat-treatment of welds has the nature of a diffusion process. It is called immunization heat-treatment or "stabilising annealing" /71/ and is rather widely used in practice.

The following facts prove the positive effect of stabilizing annealing on the corrosion resistance of welded joints made of austenitic steels.

Butt welds of 1Kh18N9T steel pipes at an oil refinery were deteriorating in the vicinity of the weld because of intergranular corrosion (Fig. 99). Stabilising annealing could eliminate such annoyances. Apparently, in the given instance, it was also quite significant that the internal stresses were eliminated as a result of heat treatment. These internal stresses contribute to the development of the so-called corrosion under stress.

Analogous facts are also known abroad. Pipes of AISI-316-type steel containing 0.11% C, 16.0% cr, 13.62% Ni, 2.95% Mo failed due to intergranular corrosion after 80,000 hours in a phenol medium at 740° F (400° C). In order to salvage some pipes which escaped failure they were subjected to heat-treatment at 870° C over a period of 24 hours. After a stabilizing annealing these pipes have effectively served for several tens of thousands of hours in an even more aggressive medium.

As we know, the σ-phase forms on austenitic-ferritic welded joints following their exposure to critical temperatures for a more or less extended period. With large quantities of ferrite the $\delta \rightarrow \sigma$ transformation can even occur during the stabilizing annealing at 850° C. The presence of primary ferrite increases the stability of the weld material against intergranular corrosion; but what will happen to the weld after the $\delta \rightarrow \sigma$ transformation has taken place?

Various data are available on the effect of the σ-phase on the stability of austenitic-ferritic steels. For example, in reference /152/ it is pointed out that while the σ-phase can provoke a drop of the over-all corrosion resistance of 18-8-type steels, in intergranular corrosion it is the carbides and not the σ-phase that play a predominant role. Shirley /209/ assumes that the effect of the σ-phase on the corrosion resistance depends on the distribution characteristics of this structural component. In single-phase steels the σ-phase forms on the boundaries of grains, thus leading to increased corrosion. In two-phase steel the σ-phase forms within the ferrite, and hence the metal does not become prone to corrosion.

Fig. 99. Intergranular corrosion destruction of 1Kh18N9T steel in heat-treated weld region: a -- external view of destruction, b -- microstructure (300X).

In view of the high chromium and molybdenum content, the σ-phase as such is corrosion resistant, but the austenite surrounding the σ-phase is depleted of these elements whereby it becomes sensitive to corrosion. In 1952 Hoare pointed out that the sigma phase somewhat increases the general corrosivity of 18-8-3-1-type (Cr-Ni-Mo-Nb) steels. Likewise, an English paper on welding austenitic, stainless and heat-resistant steels /227/ stresses the fact that the σ-phase may have a negative effect on the corrosion resistance of steels in certain media. Detailed data, however, are not given. On the other hand, Denaro and Hinde hold that the corrosion resistance of stainless steels is not negatively affected by the σ-phase. Stewart and Yurben maintain, on the contrary, that the σ-phase enhances intergranular corrosion of 18-8 steels with Ti and Al in boiling 65% nitric acid. Aluminum calls forth the formation of σ-phase not only δ-ferrite but also in austenite where it is found in a solid solution.

A. J. Lena /190/ notes that the effect of the σ-phase on intergranular corrosion of 18-8 ELC ultralow-carbon steels is noticeable only in tests by the Guillet method, i.e., in boiling concentrated nitric acid. Tests made in the Strauss-Hadfield reagent (H_2SO_4 + $CuSO_4$) do not show any intergranular corrosion caused by the precipitation of the σ-phase along the grain boundaries.

Our experiments have unequivocally proven that the $\delta \rightarrow \sigma$ transformation does not negatively affect the resistance of welds on 18-8 steels to intergranular corrosion.

Yu. I. Kasennov and others of NIKhIMMASH (Scientific Research Institute of Chemical Machinery) /36/ found that the formation of the σ-phase by means of heat treatment at 900°C imparts resistance to intergranular corrosion of stable austenitic Kh23N23M3D3 (EI533) steels. Our tests have shown that a partial formation of the σ-phase is also useful for welds of the above-mentioned steels: not only does it prevent intergranular corrosion, but it increases the over-all corrosion resistance of welded joints in boiling 75% sulfuric acid.

Mangen /193/ points out in his thesis that by maintaining 18-8 steel at very low temperatures corrosion increases by a factor of 2 to 10. According to our data, the processing of automatic welds of 1Kh18N9T steel at very low temperatures (3-hour immersion in liquid oxygen, -183°C) did not affect their resistance to intergranular corrosion, although the over-all chemical resistance in boiling nitric acid was somewhat decreased (see page 253).

7. The effect of strain hardening on the weld resistance of austenitic steels to intergranular corrosion.

It is known that cold deformation of 18-8 austenitic steels, which are not titanium- or niobium-stabilised, enhances their resistance to structural (intergranular) corrosion. As a result of cold deformation, glide planes are formed within the austenite grains, and the surface on which the excess phase can precipitate increases. Chromium impoverishment spreads to a large volume of the austenite.

At constant carbon content, the rate of austenite impoverishment becomes less "dangerous." In /173/, a very interesting example is shown of the beneficial effect of cold deformation on the resistance of stainless steels to intergranular corrosion. Bands, cut with automatic shears from a 2 mm thick 18-8 steel sheet free of titanium or niobium, were kept for two hours at 700° and boiled for 150 hours in a solution of sulfuric acid and copper sulfate. Following boiling, the entire inner part of the bands was pulverized. Only the sheared borders, subjected to strain hardening during the cutting, remained intact; instead of the plate only the "frame" remained (Fig. 100).

Actually, cold deformation has a greater effect on welds than on rolled steel. Specimens were prepared from 10 mm thick 18-8 steel sheets; these specimens were automatically welded on both sides and following shaping and polishing were stretched. Specimens' residual elongation deformation ranged from 4 to 30%. After stretching, the specimens were heated for two hours at 650°, and subsequently boiled for 72 hours in a sulfuric solution of copper sulfate. Tests have shown that the weld, after withstanding 30% stretching, is absolutely insensitive to the action of "dangerous" temperatures. As indicated in Chapter 3, cold hardening of welds leads to the appearance of secondary ferrite which precipitates near the primary ferrite regions (see Fig. 49). With the increase in the ferrite content, the diffusion rate drastically increases under the effect of critical temperatures, and the chromium impoverishment in austenite is eliminated within a minimum amount of time. In this connection it should be noted that the lesser the ferrite content in the weld the slighter is the positive effect of plastic deformation.

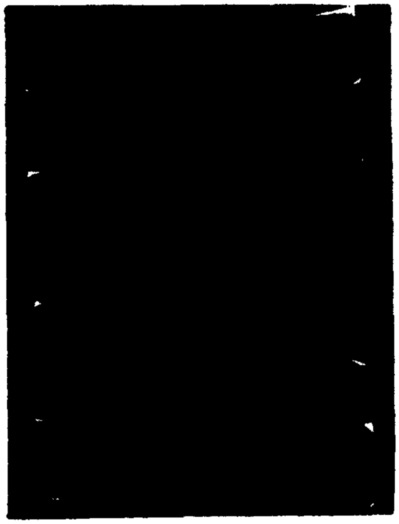

Fig. 100. The effect of cold hardening on the corrosion resistance of 18-8 rolled steel (Gudremon and Schafmeister).

8. Knife-Line Corrosion in Welded Joints of Chromium-Nickel Austenitic Steels.

As far as we know, knife-line corrosion was first mentioned in 1950 to 1952 /174, 208/. At the University of Ohio, a boiler made of 18-8 steel containing niobium was investigated; it had been rendered unserviceable by knife-line corrosion. In England, knife-line corrosion put out of service a large welded tank for concentrated nitric acid made of 18-8 stee with titanium (Fig. 101, a, b).

In 1952, together with N. A. Langer we published the first data available in the Soviet literature on knife-line corrosion of 1Kh18N9T steel specimens. At the same time assumptions were given regarding the causes of this phenomenon and the means to eliminate them. At the end of 1954, M. M. Kurtepov /45/ published a paper on knife-line corrosion of welded 18-8 steel specimens in oxidizing media. In 1955, Yu. I. Kasennov /34/ described knife-line attack in welded joints of 18-8 steel made by manual arc-welding.

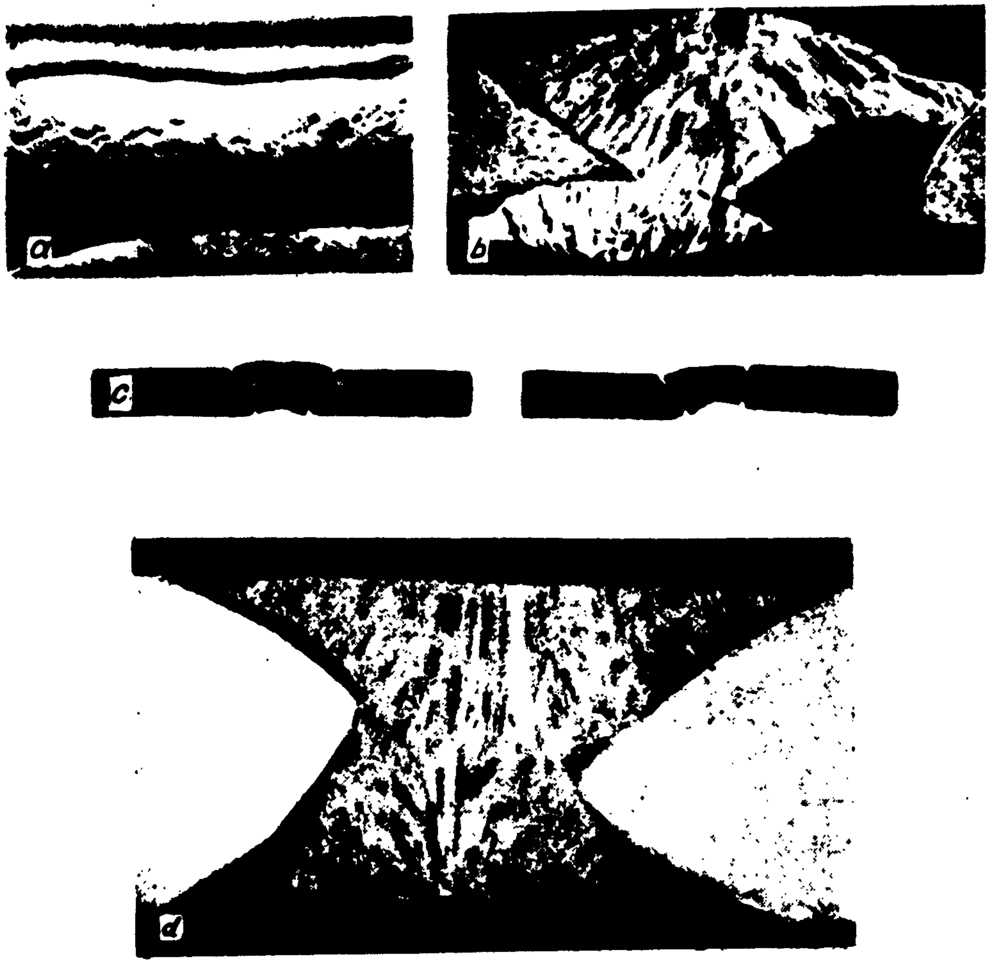

Fig. 101. Knife-line corrosion of welded joints:

a, b -- a tank of 18-8 steel with titanium, in England /208/;

c -- butt joint of 1Kh18N9T steel; d -- same of Kh18N12M2T steel.

At the present time, on the basis of data obtained at the Institute of Arc Welding between 1950 and 1958, it is possible to study the conditions and causes provoking knife-line corrosion in welded 18-8 steel joints and also to indicate the means for preventing this dangerous phenomenon /75/.

Knife-line corrosion on the boundary between weld and base metal was first discovered by us in the course of tests for intergranular corrosion of automatically welded joints of 1Kh18N9T and Kh18N11B steels. The specimens were boiled in a sulfuric solution of copper sulfate and in a nitric solution of sodium fluoride. The most typical specimens are shown in Fig. 89, c, and in Fig. 97, b. A similar phenomenon was discovered during tests on the overall chemical resistance of welded joints made with repeated and prolonged boiling in 55% and 65% nitric acid. A cross section of these specimens is shown in Fig. 101, c.

It is known that boiling in concentrated nitric acid is one of the methods for detecting intergranular corrosion of stainless steels. Hence the assumption that knife-line corrosion is a variety of the structural, i.e., intergranular, corrosion. Subsequent studies have confirmed this assumption, as will be shown hereafter.

Numerous experiments have shown that if tested in the natural state, i.e., without any heat-treatment after welding, the welded joints with single pass one-sided automatic welds seldom show knife-line corrosion. It can occur, however, if before the corrosion tests, these compounds were exposed to a more or less prolonged action of so-called "dangerous" temperatures, for example 650°C or in the case of welding thin metal plates when the cooling rate of the weld area is relatively low.

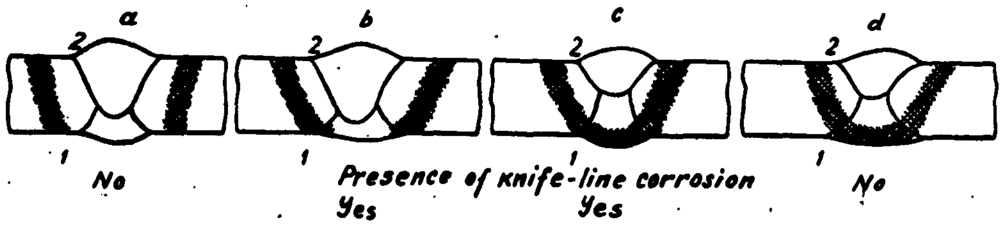

Fig. 102. Knife-line corrosion of welded specimens with two-sided welds depending on the size of the weld.

Whereas in specimens with two-sided single pass welds (Fig. 102) knife-line corrosion, as a rule, does not affect the base metal on the side of the second weld, it can be detected on the boundary of weld 1 (Fig. 102). However, if before being exposed to the action of an aggressive medium the specimen is tempered for two hours at 650°, knife-line corrosion can also occur near weld 2.

The presence or absence of knife-line corrosion near weld 1 is determined by the welding conditions and the shape of both welds, all other conditions being equal. If weld 1 is narrow and the penetration of weld 2 is sufficiently deep (Fig. 102, a), knife-line corrosion, as a rule, will not appear on the side of weld 1. However, under the same experimental conditions, knife-line corrosion can be detected if weld 1 is wide (Fig. 102, b). An analogous phenomenon occurs in those cases where weld 1 is narrow and the penetration of weld 2 is comparatively shallow (Fig. 102, c). On the other hand, there will be no knife-line corrosion if weld 1 is wide and the penetration depth of weld 2 is small (Fig. 102, d). In other words, knife-line corrosion can only occur if during the welding of weld 2 portions of the base metal directly adjacent to weld 1 are exposed to the action of "dangerous" temperatures.

Knife-line corrosion, however, does not only occur in such oxidizing media as nitric acid. Since this type of failure by corrosion is observed in the Strauss-Hadfield reagent (10% H_2SO_4 + 10% $CuSO_4$), we could also expect it to appear after processing specimens in boiling non-oxidizing media, such as sulfuric acid.

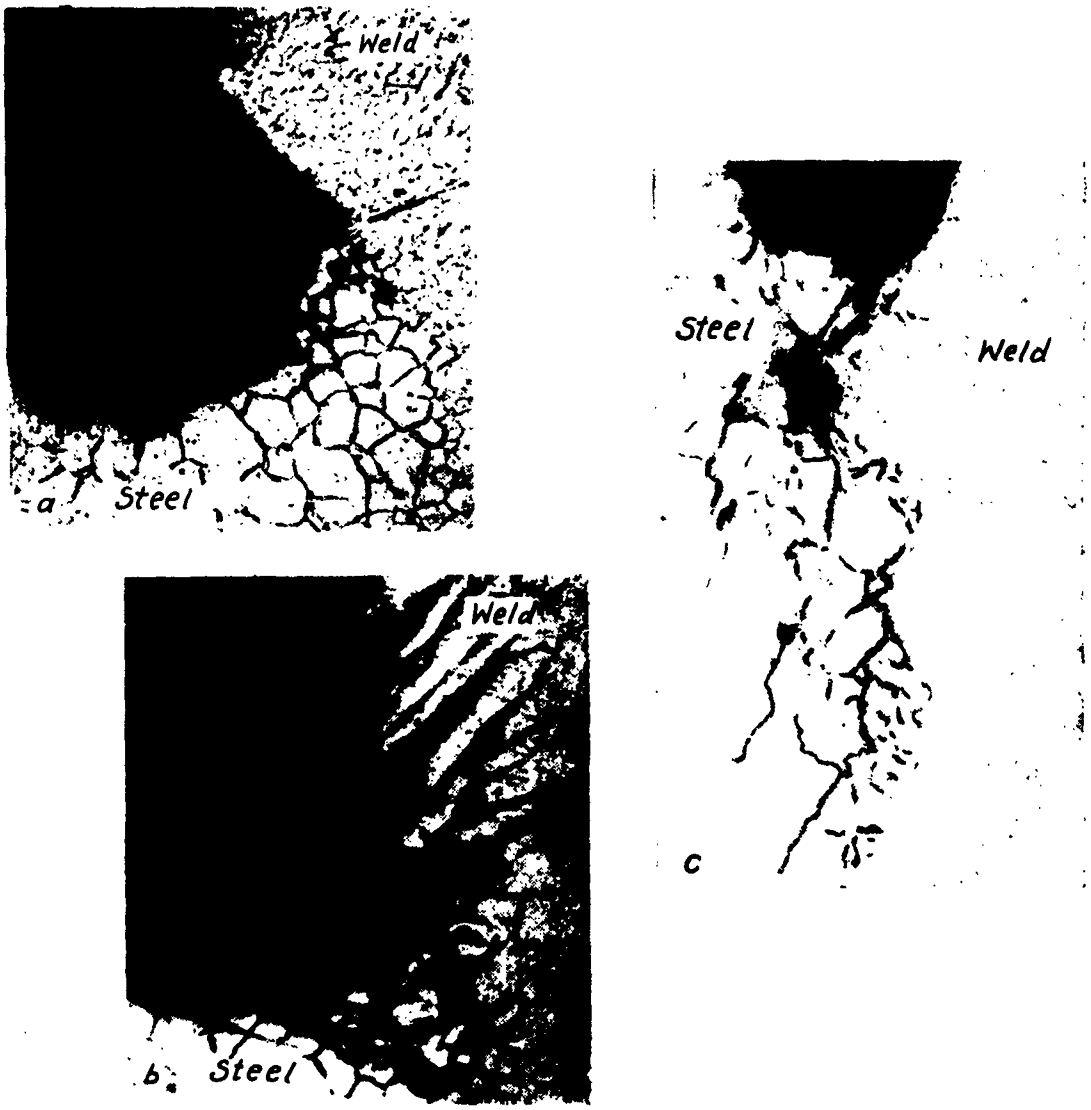

Fig. 103. Knife-line corrosion as a variety of intergranular corrosion: a- welded joint of 1Kh18N9T-type steel after boiling 250 hrs in 55% nitric acid (130 X); Б - same, weld with austenitic structure (200 X); c - same, weld with austenitic-ferritic structure (130 X).

Experiments have confirmed this assumption. In fact, in a sulfuric medium intensive knife-line corrosion affects not only 1Kh18N9T- and Kh18N11B-type steels, but also Cr-Ni-MoTi Kh18N12M2T-type steels having an elevated chemical resistance in this reagent.

As previously mentioned, metallographic investigations have shown that knife-line corrosion is of an intergranular nature (Fig. 103). By deeply penetrating into the base metal the aggressive liquid first affects a very narrow strip of metal (not wider than 3 to 5 grains) immediately adjacent to the weld. Then, the front of knife-line corrosion somewhat widens as the intergranular layers are progressively destroyed, the bond between the grains is disrupted and the grains are washed out. However, the width of this front never exceeded 1 to 1.5 mm in any of the many dozens of specimens investigated by us.

Owing to concentration, the corrosion process is disastrously accelerated. In one case it proceeded at such rate that after 24 hours of boiling in a 6% solution of H_2SO_4 a welded sepcimen, 10 mm thick, almost completely failed in its entire thickness: in 24 hours the reagent had penetrated into the stainless steel to a depth of approximately 6 mm (Fig. 101, d), which corresponds to an unprecedented rate of corrosion, viz., over 2 km per year. This, and many other examples, clearly indicate that knife-line corrosion is one of the most alarming forms of failure in stainless steels.

The development of knife-line corrosion is closely related to the structure of the adjoining sections of the weld and the base metal. In psecimens with purely austenitic welds knife-line corrosion of the base metal is followed by a fairly noticeable failure of the adjacent welded joint (Fig. 103, b). In austenitic-ferritic weld specimens (Fig. 103, a, c) there is no failure of the weld metal proper. In this connection it must be stressed that knife-line corrosion depends only to a negligible degree on the size of the steel grains in the affected area. Indeed, it can equally develop in fine-grained and in coarse-grained austenitic steels.

An analysis of the conditions producing knife-line corrosion permits of the indisputable conclusion that this kind of failure can affect those portions of stainless steel which were first subjected to heating at temperatures close to the melting point and subsequently exposed to a more or less prolonged action of so-called "dangerous" temperatures (for example, 650°).

Let us now establish what happens to 18-8 steels stabilized with titanium or niobium + tantalum after a heat treatment of this kind. In the final state, i. e., after quenching for austenite from temperatures ranging between 1050 and 1100°C, at room temperature these steels usually possess an austenitic structure or, depending on their chemical composition, sometimes also an austenitic-ferritic structure with a flow-line distribution of the ferritic component (cf. Fig. 6, b). Carbon is partly found in a solid solution (up to 0.02-0.03%), its bulk (above 0.02-0.03%) being combined in titanium or niobium and tantalum carbides evenly distributed in the austenite where, as a rule, some small quantities of Ti and Nb + Ta are also found in a solid solution. If this steel is heated for 1 lo 2 hours at 600-700°C, its over-all corrosion resistance could sharply drop by a factor of ten and more(cf. page 250), but intergranular corrosion will not appear since there is no chromium impoverishment of the austenite grain boundaries. (It should be noted that the causes for the drop in the over-all corrosion resistance of stabilised steels subjected to tempering have not been sufficiently studied as yet.)

During heating of stabilized steels at high temperatures (above 1300°C), the titanium and niobium + tantalum carbides begin to dissolve. Carbon, becoming free of these elements, penetrates the austenite lattice, while titanium and niobium + tantalum occupy the lattice vacancies.

Diffusion processes are sharply accelerated with rising temperatures. Hence carbon atoms, with an incomparably greater mobility than titanium or niobium atoms, leave the inner regions of austenite grains, where they had been located earlier in the form of Ti and Nb + Ta carbides, rapidly escape toward the peripheries of these grains, while remaining, of course, within the interstitial solid solution. This is probably enhanced by the horophile property of carbon with regard to chromium-nickel austenite /8/. Rapid cooling of steel heated to temperatures above the solubility point of carbides, will fix the austenitic structure [redacted]. The austenite thus obtained contains stabilizing elements (Ti, Nb + Ta) in the substitute solid solution, mainly within the grains. The austenite is distinguished by a higher carbon concentration on the grain boundaries, the carbon being found in a supersaturated solid solution. During subsequent tempering (heating at 650 to 700°C), the first carbon atoms precipitating from the solid solution along the boundaries of austenite [redacted] will capture all the titanium atoms in the boundary layers, and [redacted] form titanium carbides. The new carbon atoms, however, diffusing from the inner regions of austenite grains toward the peripheries, will not find a sufficient quantity of titanium atoms. Hence, there will be chromium carbides rather than titanium carbides forming on the grain boundaries. This is due to the fact that carbon diffuses much faster in austenite than chromium, the diffusion rate of the latter being higher than that of titanium. Hence, 18-8 steels containing titanium may become prone to intergranular corrosion. Monypenny /192/ gives the following example. Steel quenched from 1050 to 1100°C withstands the ensuing 30-minutes heat treatment at 650°C without becoming susceptible to intergranular corrosion. The same steel quenched from 1250°C becomes susceptible to intergranular corrosion after a 5-minute exposure to 650°. Morley's paper contains the same data. According to N. S. Alferova and L. P. Shchesno /5/ superheating brings about intergranular corrosion in 1Kh18N9T-type steels. Brown has found that by raising the quenching temperature of 18-8 steels containing titanium to 1200°C and higher, intergranular corrosion appears after tempering at 650°. It can be prevented by raising the titanium content to fulfill the condition $\frac{Ti}{C} \geqslant 18$ rather than $\frac{Ti}{C} \geqslant 6$, the latter being essential for steels quenched from temperatures of not more than 1100°C. Mangen /193/, too, stresses that an increase in the quenching temperatures from 1050 to 1180°C will intensify intergranular corrosion.

According to M. I. Vinograd's data, 18-8 steels containing Ti or Nb + Ta quenched from 1300°C become affected by intergranular corrosion only after tempering at 650°C(*).

The above-described pattern of transformation of stabilized steel (i.e., resistant to intergranular corrosion) into corrodible steel explains the causes for knife-line corrosion. Yet, there is no explanation for the fact that not all welded joints are attacked by it, although in many, if not in all, cases the temperature conditions are those required for knife-line corrosion. This may be attributed to the different composition and structure of the steel in question.

If the titanium and niobium + tantalum content of steel exceeds that required to bind carbon into carbides, or if the carbon content is low, knife-life corrosion may not occur. In this case, the atoms of the stabilizing elements found in a solid solution near the grain boundaries of austenite, during tempering combine all the carbon precipitating from the solid solution in the superheated section of the weld area into carbides, thus preventing the chromium depletion of austenite.

References /174, 208/ point out that steels with an increased Ti and Nb + Ta content are affected by knife-line corrosion only to a neglible degree.

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[*] Private communication.

Our tests show that two-phase steels with a flow-line distribution of the primary ferrite are more resistant to knife-line corrosion than single-phase austenitic steels. In the former case, the ferrite strips present an obstacle which is hard to overcome by the aggressive reagent on its way deep into the steel (Fig. 104).

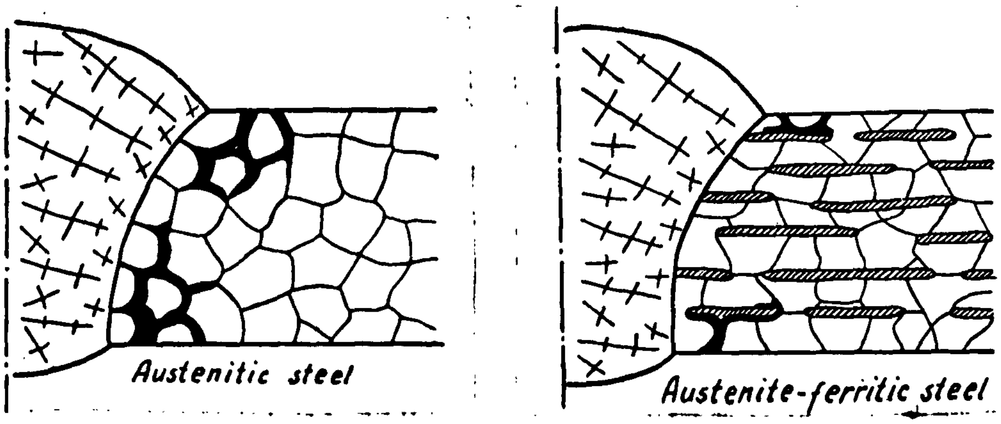

Fig. 104. Diagram of the development of knife-line corrosion in austenitic and austenitic-ferritic steel.

Above, we examined the main cause of knife-line corrosion, i.e., the dissolution of Ti and Nb carbides and the subsequent formation of chromium carbides. There are, however, a number of other factors which play a part in making steel susceptible to knife-line corrosion. Among them are: an increase in the intragranular inhomogeneity of austenite in the area immediately adjecent to the weld, and the diffusion of carbon and other elements from the steel into the weld pool.

It has recently been established that in spite of short-enduring high-temperature heating of the weld region, arc welding could produce carbon diffusion from the steel into the weld pool during its crystallization. Hence one can easily assume that carbon diffusion from the weld area into the weld, for example, in the case of welding 18-8-type steels with extra-low-carbon steel electrodes, can prevent knife-line corrosion. Conversely, a faster diffusion of titanium, and niobium + tantalum from the steel into the weld, made possible by an equal carbon concentration in the steel and in the pool, may promote the formation of knife-line corrosion.

It follows from what has been said above that the following measures for preventing knife-line corrosion may prove of some use: 1) averting the precipitation of chromium carbides and the chromium impoverishment of the austenite grain boundaries, and 2) elimination of chromium impoverishment of the austenite.

In order to prevent the chromium carbides from precipitating in the superheated weld section it is necessary:

a) to raise the Ti and Nb content of steel above the limits set by Formulas (2) and (3). We mentioned above that many researchers have of late been advancing this requirement for obtaining a reliable means for preventing ordinary intergranular corrosion in these steels;

b) to use two-phase steels with an austenitic-ferritic structure. Unfortunately this recommendation cannot be applied in the case of 18-8 steels in non-oxidizing media, since here the presence of the ferritic component engenders an abrupt drop in the over-all corrosion resistance of steels (see page 246);

c) to employ such welding rogimes and techniques as to prevent the portions of the weld section facing the aggressive medium from coming within the range of action of the critical temperatures during the welding of weld #2 (see Fig. 102);

d) to use steels with a minimum carbon content.

If precipitation of chromium carbides in the above-mentioned area cannot be prevented, appropriate measures should be taken to eliminate chromium depletion of the austenite grain boundaries caused by this precipitation. Entirely reliable results in this respect are obtained by a stabilizing anneal of the welded joint (heating for 2 or 3 hours at 850-900°, and subsequent cooling in air. This heat treatment of 18-8 steels containing titanium of niobium leads to a stabilization of the carbides of these elements. According to M. B. Shapiro and G. L. Shvarts, the most complete combination of carbon into titanium carbides occurs after heating 1Kh18N9T-type steels at 850 to 900°. If during quenching from 100 to 1150°C the titanium content of the carbides is only 0.15%, then after quenching from these temperatures a maximum titanium content (in the given case, 0.50%) will be detected in the carbides.

Owing to the fact that knife-line corrosion in weldments of 18-8 steels containing titanium is becoming more frquent, special investigations were undertaken in England for finding means to prevent this phenomenon. In a paper published in 1956 Heeley and Little /179/ held that a radical measure against knife-line corrosion is rigid limitation of the carbon content (not to exceed 0.06%) in 18-8 steels containing titanium. Such a measure would also prove useful from the viewpoint of increasing the over-all corrosion resistance of steels and welded joints. These authors maintain that a stabilizing annealing prior to welding will not prevent knife-line corrosion if $Ti/C \leqslant 5.5$, but may prove helpful if $Ti/C > 7$.

9. Resistance of Welded Joints of Austenitic Steels to Over-all Corrosion Oxidizing and Non-Oxidizing Liquid Media.

It is known that equipment made of acidproof chromium-nickel austenitic steel is used in many branches of industry. It is doing service in various aggressive media. Therefore, the behavior of a welded stainless steel joint in any one medium cannot, strictly speaking, characterize the resistance of this joint under some other service condition. Hence the results of any laboratory-type tests on over-all corrosion, regardless of the aggressive medium employed, are always somewhat hypothetical. For this reason, considerable attention has lately been paid to corrosion tests made under working conditions. Judging from literature, it seems to happen frequently that specimens which did not withstand laboratory tests displayed excellent behavior under actual service conditions. Incidentally, analogous data are also available for resistance of weldments to intergranular corrosion /214/. Nevertheless, laboratory-type tests for over-all corrosion are indispensable, since they allow to make at least a relative evaluation of the quality of weldments.

At present, there is no standardized research procedure for testing the over-all chemical resistance of weldments of austenitic chromiun-nickel steels. The most common research method is treating specimens in boiling nitric acid of different concentrations, whereby the corrosion resistance can be determined from the loss in weight resulting from several boiling cycles, each cycle lasting for 24 to 48 hours. The size of the welded specimens has a considerable bearing on the results of these tests. In laboratory conditions tests with large specimens are difficult to perform. As a rule, their size has to be considerably reduced. However, the welds will display a different behavior in one and the same aggressive medium with a different ~~steel surface-to-weld~~ steel-to-weld surface ratio if under actual service conditions the weld will serve as an anode with respect to the base metal. Hence it is desirable that this ratio be as high as possible.

According to GOST 5272-50 "Corrosion of metals, terminology", the extent of corrosion is usually determined by the specific weight loss in unit time (g/m^2 hour). Abroad, the extent of corrosion is usually determined by the thinning rate of the material (mm/year). As mentioned, the results of determining the absolute loss in weight of the specimen after boiling were in both cases taken as initial data.

The method of evaluating the corrosion resistance of welded joints by the over-all loss in weight does not take account of a possible local concentration of corrosion on the weld or in the adjoining area. Hence it would be desirable to apply the method of measuring the penetration depth of corrosion, developed some time ago by D. M. Rabkin and others at the Institute of Electrical Welding, to welded joints of carbon and low-alloy steels. This method requires a basis for measuring the profile of the metal surface, ~~profile~~ which varies after each series of tests. The boiling nitric or sulfuric acid used for corrosion tests of stainless steels is so aggressive that the formation of absolutely acidproof measuring bases is a highly complex task*. Owing to this, the method of determining the corrosion resistance of welded joints of austenitic steels by weighing remains, for the time being, the only reliable one. In the experiments effected by us together with N. A. Langer we also made use of visual inspection and qualitative evaluation of the corrosion intensity ~~of the~~ in various sections of

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[*] A section of the weld area heated during the welding process to temperatures up to 1050-1100°C and not subjected to the subsequent action of ~~the~~ "dangerous" temperatures, possesses maximum corrosion resistance. N. Yu. Pal'chuk proposed therefore to use it as a basis for measuring the penetration depth of corrosion into various sections of welded joints.

welded joints. In addition, after the corrosion tests we proceeded in several cases examine the transversal templets cut from the welded specimens in order to establish the penetration depth of localized corrosion /67, 68, 71/ (cf. Fig. 101, c).

Over-all corrosion by liquids, like intergranular corrosion, is an electro-chemical process. It should be borne in mind that the over-all chemical resistance of the weld metal does not only depend on whether the weld is an anode or a cathode with respect to the base metal, i. e., on the "macroconditions" for the development of corrosion, but also on the "microconditions" under which the corrosion processes take place on the weld. Hence the structural conditions of the weld, the degree and character of its alloyage, the presence or absence of phase transformations, the rate of chromiun depletion of the austenite boundaries — in short, all those factors bearing upon the failure by intergranular corrosion must also be of importance for the over-all chemical resistance of the weld metal.

A simple example evidences the correctness of this assertion. In a welded specimen with a double butt weld, ~~there appears a concentrated corrosion~~ localized corrosion appears on weld 1 subjected to repeated heat treatment during the welding of weld 2 (Fig. 105).

The strongest failure was observed in that section of weld 1 ~~being~~ which is at a certain distance from weld 2. It can easily be seen that the shape and location of this area is determined by isothermal surfaces responding to the critical ("dangerous") temperatures. In the immediate vicinity of weld 2, the metal of weld 1 was heated to temperatures at which the excess phase dissolves. Here, corrosion develops normally. During the welding of weld 2, the metal of weld 1 was heated to a lower temperature at a certain distance from weld 2. In spite of the short action of the critical temperature, the excess phase precipitated here, thus leading to an intensification in the over-all corrosion. Let us now examine in greater detail the effect of the "microconditions" on the over-all corrosion resistance of welds of stainless chromium-nickel steels.

The effect of the "microstructure." Until recently, the problem of the effect of the microstructure of 18-8 steel welds on their over-all corrosion resistance did not attract sufficient attention. Nevertheless, quite a few papers were published on this subject with a particular regard, however, to the base metal. Various viewpoints were expressed. Referring to Uhlig, Wetternick notes that in halogenous solutions austenitic steels are stronger affected by corrosion than austenite-ferritic steels. (The addition of small quantities of silver to austenitic steels somewhat increases their resistance to these solutions).

Shirley and Nicholsen /208/ maintain that two-phase steels are inferior to single-phase steels with respect to their resistance to nitric acid. In two-phase steels failure by corrosion proceeds along the boundary line between austenite and ferrite.

When the new American standard for welding electrodes for austenitic steels providing for an increased ferrite content in the weld was introduced in 1954, it was noted that both ferrite and austenite are equally corrosion resistant in oxidizing media. In this case the corrosion resistance results from the formation of a film of chromium oxides, and this depends on the chromium concentration in the metal regardless of the latter being austenitic or ferritic /215/.

D. O. Slavin /122/ points out that up to 20% of ferrite in 1Kh18N9T-type steels virtually has no bearing on their corrodibility.

In reference /182/ it is pointed out that although there are single examples of predominantly corrosive failure of austenite or ferrite, steels with a single-phase (γ) and a two-phase ($\gamma + \delta$) microstructure practically have an equivalent over-all corrosion resistance.

Fig. 105. Corrosion failure of a welded specimen with a double weld.

Many authors believe that the presence of the δ-phase in steel accelerates the over-all corrosion in the nitric acid. According to Thielsch and Pratt /219/, the δ-phase and the copper-rich ε-phase, which is formed in high-alloy 25-20 steels with copper and molybdenum, can appreciably decrease the corrosion resistance of these steels in sulfuric acid. Our tests of welding Kh23N23M3D3 steels provide no proof for this conclusion.

The effect of the microstructure of 18-8 steel welds on their resistance to over-all corrosion in oxidizing media can be judged from the results of the following experiments.

Two series of 12 mm thick automatically welded 1Kh18N9T steel specimens were tested for corrosion in boiling 65% nitric acid. The dimensions of the specimens were 40 x 120 mm. In one specimen series the welds had an austenitic-ferrite structure (up to 10% δ-phase); in the other, the welds had an austenitic structure owing to higher nickel content.

These corrosion tests have shown that regardless of the heat-treatment the austenitic-ferrite structure welds as well as the austenitic structure welds have an equal over-all loss of weight. However, an examination of surfaces of the specimens subjected to tempering reveals a different picture: whereas, the corrosion of an austenitic ferrite weld is uniform, the surface of the austenitic weld is heavily ulcerated due to local corrosion concentration.. Furthermore, in welded joints of austenitic welds we observe a high corrosion concentration in the weld region. For specimens of austenitic welds subjected to tempering, the depth of corrosion penetration is 1 mm in 120 hours, which is equivalent to 73 mm/year. Should corrosion progress uniformly, the depth of corrosion penetration, at an over-all loss of weight of 7.93 gr/m^2 hr, would be only about 8 mm/year, and that means 9 times less. Intense knife-line corrosion in specimens with purely austenitic welds can be explained by the fact that, here, beside the superheated region of the base metal the weld metal is also subject to intense corrosion (see Section 8, Fig. 103, b).

Experimental data show that austenitic-ferrite welds (of a δ-phase content up to 15 to 20%), far from being inferior to austenitic welds, have an even higher resistance to over-all corrosion in oxidizing media, especially after being tempered at 650°C. This situation proved to be somewhat paradoxical: the uniform, homogeneous austenitic structure proves to be less chemically stable than a nonuniform heterogeneous two-phase structure. It should be kept in mind however that the concept of the so-called uniformity of austenitic welds is a false concept. Due to the peculiarity of the primary transcrystalline structure with thickened intergranular layers, these welds are actually less uniform than those which have a disoriented and granulated austenitic-ferrite structure with thinned and intercrystalline layers dissociated by primary ferrite. This specific feature of the primary structure of the austenitic-ferrite welds which provides for their high resistance to intergranular corrosion, also accounts for their high resistance to over-all corrosion in oxidizing media. Metallographic investigation, made following the tests for over-all corrosion in 55 to 65% nitric acid illustrates that in a purely austenitic weld, destruction caused by corrosion occurs mainly along the boundaries of acicular crystals (Fig. 106, a). In austenitic-ferritic welds, corrosion is uniform both at the grain boundaries and in the austenite crystals containing primary ferrite regions (Fig. 106, b). Analogous data are listed in /210/.

Two-phase welds lose the above advantages as regards their resistance to over-all corrosion in nitric acid (oxidizing medium)

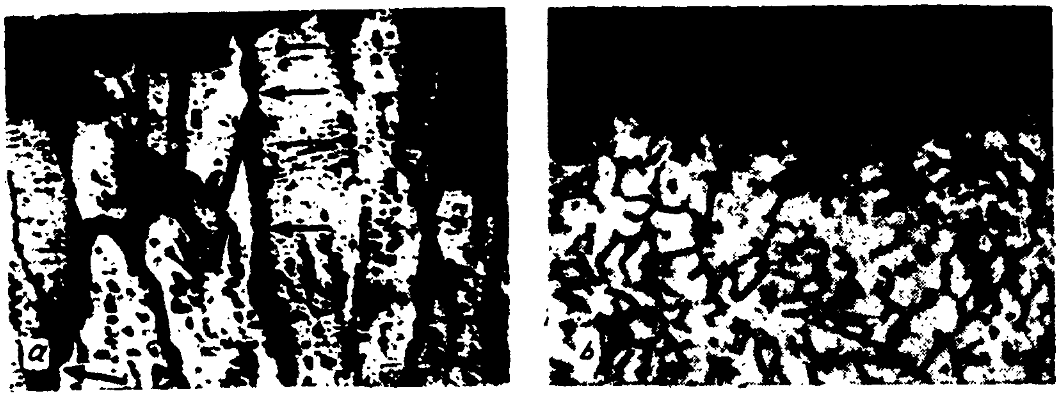

Fig. 106 . Effect of the microstructure of a weld on the character of the development of over-all corrosion: a -- austenitic weld, with preponderant destruction of boundaries of acicular crystals (60X); b -- austenitic-ferrite weld -- uniform corrosion of both phases (200X).

if sulfuric acid (non-oxidixing medium) is used instead. This, possibly, is linked to the low resistance of chromium-ferrite in non-oxidizing media. It is well known that the chemical stability of austenitic chromium-nickel 18-8 steels and that of 17% chromium steels in boiling nitric acid is practically equal. Here, in an oxidizing medium, the resistance of a steel is determined by its passivation capacity /44, 129, 130/, i.e., it depends entirely on the chromium content of the steel. This is different in sulfuric-acid media, in which the resistance of ferrite steels of equal chromium content is incomparably lower than that of austenitic steels. The above-said can be illustrated by the following example. Specimens of chromium-ferrite Kh17 (17% chromium and 0.05% carbon content) steel with welds containing 18% Cr and 8% Ni were boiled for 24 hours in 65% nitric acid as well as in 10% sulfuric acid. The specimens boiled in nitric acid preserved their external appearance (Fig. 107, a). In specimens boiled in sulfuric acid , only the austenitic welds remained; the ferrite steel which was the base metal -- dissolved completely (Fig. 107, b).

The lower the percentage of ferrite in welds and 18-8 steels, the higher their resistance to sulfuric acid. However, 18-8 steels, even such of purely austenitic microstructure, are insufficiently resistant to sulfuric acid. Austenite-stable steels of an increased chromium and nickel content, and with additional admixtures of copper and molybdenum, are reliable in the media in question (see Table 2).

Fig. 107. Welded specimens of Kh17 steel with welds containing 18% Cr and 8% Ni, after boiling: a) in nitric acid, b) in sulfuric acid.

Due to their purely austenitic structure, welds of such steels show a resistance to over-all corrosion not inferior to that of the base metal. Thus, for example, after 100 hrs boiling in sulfuric acid of different concentrations (35, 50 and 75 per cent), the loss in weight of specimens of automatically welded Kh23N23M3D3 (EI533) steel was just as small as the loss of the specimens of the (unwelded) base metal. As a result of tempering at 650 to 700 C, the resistance to over-all corrosion of these weld joints in sulfuric acid decreases slightly.

It has been established that the corrosion resistance of 18-8 steels in sulfuric acid of low concentration can be cosiderably increased by way of small additions of certain substances to the aggressive medium. The following relevant data are known to us.

Kiefer and Renshaw found that the addition of various sulfates, such as $CuSO_4$, $SnSO_4$, $NiSO_4$, $MnSO_4$, $FeSO_4$, $(NH_4)_2SO_4$ to sulfuric acid moderates corrosion in chromium-nickel austenitic steels. Most effective are strong oxidizing agents, such as nitric acid, chromic acid, and dichromates (for instance, $K_2Cr_2O_7$), which entirely prevents [illegible] corrosion of austenitic steels in sulfuric acid.

Kh. L. Tseytlin contends that small concentrations of haloid ions have a moderating (inhibiting) effect on the corrosion of 18-8 steels in 10N sulfuric acid. The presence of 0.3 to 0.6% of the chlorine ion in the solution, reduces corrosion of 18-8 steel (by) 40 to 60 times. Our tests with weld specimens did not confirm these data. As regards oxidizers in small additions, they do actually retard corrosion of 18-8-steel welds in sulfuric acid (Table 48).

Table 48

Effect of HNO_3 (oxidizer) addition to sulfuric acid on the corrosion resistance of welds of 1Kh18N9T steel

Brand of electrode used in welding	Loss of weight (in g/m^2hr) of weld specimens afer boiling in 6% H_2SO_4 with addition of HNO_3			
	None	0.1%	0.3%	0.5%
Sv-OKh18N9	33.3	25.7	0.32	0.10
Sv-OKh18N9S2	53.3	48.0	0.56	0.15
OKh18N9FBS (EI649)	46.3	34.9	0.68	0.12

In their work Thielsch and Pratt cite data according to which the addition of a mere 0.1% of the strong oxidizer $Na_2Cr_2O_7$ to 10% H_2SO_4 retards the corrosion of 20-23 steel with 3% molybdenum and 2% copper by a factor of 70 /219/.

In the work of Uhlig it is shown that small additions of strong oxidizers (HNO_3) sharply moderate the corrosion of austenitic 18-8 steels in sulfuric acid*.

Other data show that the blowing of sulfuric acid with air (oxygen) has the same moderating effect on the corrosion of 18-8 steel as the addition of oxidizers.

In collaboration with N. A. Langer we tested weld specimens of 1Kh18N9T steel with automatic welds for corrosion in 6% sulfuric acid with and without addition of a strong oxidizing agent, in our case, of nitric acid. The results of these tests as given in Table 48 show that the corrosion of weld joints in dilute sulfuric acid can be sharply retarded (hundreds of times) by the addition of a mere 0.3 to 0.5% of HNO_3. Our experiments showed that the addition to sulfuric acid of $NiSO_4$, $MnSO_4$, $FeSO_4$, $(NH_4)_2SO_4$ and $Al_2(SO_4)_3$ as well as NaCl practically does not moderate the corrosion in 18-8 steel welds.

Aeration of sulfuric acid solutions (blowing with air) also has a great effect on moderation of corrosion in weld specimens of 1Kh18N9T steel (Table 49).

Work /44/ was the first to establish the low corrosion resistance of stainless steels in oxidizing solutions under certain conditions.

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*Corrosion of Metals, Vol. 1, Goskhimizdat, 1952.

Table 49

Effect of aeration on the diffusion rate of weld specimens of 1Kh18N9T steel in boiling 6% sulfuric acid

Brand of electrode	Conditions of corrosion tests	Loss of weight after boiling in g/m^{2}hr			
		Cycle I 4 hours	Cycle II 6 hours	Cycle III 8 hours	Cycle 4 hrs
Sv-0Kh18N9*	Normal	122.0	83.3	112.5	80.3
	Blown with air	68.2	0.67	0.25	none
0Kh18N9FBS** (EI649)	Normal	161.0	127.0	94.5	116.5
	Blown with air	84.0	1.4	2.6	none

* weld containing 3 to 5% δ-ferrite

** weld containing 10 to 15% δ-ferrite.

This phenomenon has been called "repassivation" or "transpassivity". N. D. Tomashov and G. P. Chernova /129, 130/ have shown that the increase in the rate of corrosion of stainless steels in strong oxidizing solutions, is caused by the formation of ~~the~~ soluble oxides of hexavalent chromium in the form of CrO_3 or as $(Cr_2O_7)^{2-}$ anions. De Long points out that chromium ions added to 18-8 steel in an amount of 0.003%, or to 18-8 steel with molybdenum in an amount of 0.006%, cause an abrupt increase in the corrosion rate in nitric acid.

In collaboration with N. A. Langer and M. M. Kurtepov, we carried out experiments to compare the corrosion of the weld joints of three acid-resistant steels in oxidizing media. Table 50 gives the chemical composition of these automatic welds, while the results of the corrosion tests are shown in Table 51.

Table 51

Chemical composition of the weld metal of three acid-resistant austenitic 18-8 steels

Brands of steel	Brands of electrodes	Chemical composition of weld metal in %								
		C	Si	Mn	Cr	Ni	Ti	Nb	Mo	V
Kh18N9T	Sv-OKh18N9	0.09	0.82	0.70	17.8	9.2	0.16	—	—	—
	OKh18N9FBS (EI649)	0.08	1.23	0.80	18.2	9.8	0.27	0.32	—	1.0
Kh18N11B	Sv-1Kh18N9B	0.09	1.26	1.0	18.3	11.0	0.10	1.10	—	—
	OKh18N9F2S (EI606)	0.08	1.33	1.0	17.5	11.5	0.10	0.89	—	0.89
Kh18N12M2T	Sv-Kh18N11M	0.09	0.91	0.97	17.0	12.3	0.25	—	236	—

Table 51

Corrosion resistance of weld joints of acid-resistant 18-8 steels in oxidizing media

Brand of steel	Brand of electrode	System of additional alloying of weld metal	Loss of weight of weld specimens after 200-hr boiling in solutions in g/m^2hr		
			15% HNO_3	15% HNO_3 - 10% $K_2Cr_2O_7$	55% HNO_3
1Kh18N9T	Sv-0Kh18N9 (EI649)	none Si, V, Nb	0.05 0.03	1.65 1.18	0.87 0.35
1Kh18N11B	Sv-1Kh18N11B (EI606)	none V	0.02 0.04	2.15 2.79	0.25 0.34
Kh18N12M2T	Sv-Kh18N11M	none	0.02	1.59	0.79

Above data show that the addition of the strong oxidizer $K_2Cr_2O_7$ to a relatively weak solution of nitric acid causes corrosion rate increase by more than one hundred times. A higher concentration of nitric acid is far less effective.

It has been established that, inspite of causing a multifold increase in the overall corrosion of weld joints, the addition of $K_2Cr_2O_7$, at the same time, does not cause any appreciable increase in the rate of intergranular corrosion. Thus, for example, at a relatively low rate of overall corrosion in 55 per cent HNO_3

(0.35 g/m^2hr), the penetration depth of knife-line corrosion in welds of 1Kh18N9T steel made with an 0Kh18N9FBS electrode, was a mere 0.8 to 0.9 mm for 200 hrs. After boiling in a solution of 15 per cent HNO_3 and $K_2Cr_2O_7$, the penetration of knife-line corrosion was a mere 0.15 to 0,25 mm, despite the fact that overall corrosion progresses, here, three times as fast (1.18 g/m^2hr).

Effects of heat treatment. Corrosion tests to study the effects of heat treatment showed that:

1. After quenching for austenite, if quenching temperature does not exceed 1050 to 1150°C, the resistance to overall corrosion of weld joints is the same as after welding (where there is no heat treatment);

2. After exposure to critical temperatures, the resistance of the weld joints in oxidizing media diminishes appreciably, regardless of the structure of a weld. It should be noted that tempering at 650°C affects the 1Kh18N9T steel even stronger than the weld joints. In separate cases, we observed a tenfold increase in corrosion of 1Kh18N9T steel after tempering for 2 hrs at 650°C.

Keating's data are analogous /182/. He states that tempering after quenching for austenite has no effect on the corrosion resistance of steel either in 0.5% HCl or in 1.5% H_2SO_4. In nitric acid the tempering increases the corrosion of 18-8 steel with titanium (Table 52).

If the weld joints have been quenched for austenite the negative effect of critical temperatures on the chemical stability of the weld specimens increases sharply. It should be born in mind that

Effect of tempering at 650 °C on the corrosion resistance of 18-8 steel in 70 per cent HNO_3 /182/

(corrosion rate after holding for 29 days in nitric acid in mm/year)

Heat treatment prior to corrosion tests									
Quenching from 1050 °C and tempering at 650 °C in hrs					Quenching from 1250 °C and tempering at 650 °C in hrs				
none	0.5	4	48	672	none	0.5	4	48	672
0.285	0.331	1.35	2.42	4.51	0.388	4.25	5.88	1.9	13.7

a quite similar increase in the detrimental effect of tempering after quenching is observed also in the tests for intergranular corrosion /see page 230/. From this, it may be concluded that both the resistance to over-all corrosion in an oxidizing medium, and the resistance to intergranular corrosion of a weld, are associated with its microstructure, the character of excess-phase precipitation and the dislocation of the chromium-poor regions of the austenite.

The negative effect of critical temperatures on the resistance to over-all corrosion of welds in oxidizing media becomes the more apparent the higher the quenching temperature. A quenching temperature of more than 1150 to 1200 °C if followed by exposure to critical temperatures, appreciably increases the **over-all** corrosion of weld joints in boiling nitric acid. We know the analogous effect of quenching temperature on the resistance of a true austenitic steel (Table 53) /151/.

Table 53

Effect of quenching temperature on the corrosion resistance of niobium-bearing 18-8 steel in 65 per cent boiling HNO_3 /151/

Tempering after quenching in the corrosion tests	Corrosion rate (inch·10^{-3}/year) after quenching from temperatures in °C					
	980	1090	1200	1320	1350	1370
None	23.1	24.8	24.5	22.7	22.3	20.8
for 2 hrs at 650°C	42	41	42	84	367	690

We see that heating at 1320 to 1370°C followed by tempering causes a multifold increase in the corrosion rate. This is the explanation, as has been said in Section 8, for knife-line corrosion forming in the welds of steels stabilized with titanium or niobium.

The above-mentioned conclusions in regard to the effect of heat treatment on the **over-all** chemical stability of weld joints were confirmed in full extent by the tests conducted with austenitic-ferritic welds, alloyed additionally with the following ferritizing elements:

titanium, niobium, chromium, aluminum, silicon, vanadium and tungsten.

Considering the above, it might well be assumed that heat treatment for immunization, and, in particular, stabilizing annealing, must have a beneficial effect on the resistance to over-all corrosion of welds. The tests we made in collaboration with N. A. Langer /71/ confirmed this assumption and showed that stabilizing annealing is an effective method for increasing the over-all chemical resistance of weld joints both in oxidizing and non-oxidizing media /Table 54/.

TABLE 54. Effect of Stabilizing Annealing of Weld Specimens of 1Kh18N9T Steel on their Corrosion Resistance in Boiling 65% Nitric Acid after Four 24-Hr. Cycles of Boiling (Loss of Weight in g/m^{2}hr)

Heat treatment after annealing	Heat treatment after welding			
	none	annealing for 2 hrs. at 900°C	annealing for 2 hrs. at 850°C	annealing for 4 hrs. at 800°C
none................	2.2	1.9	1.8	1.7
2 hrs at 650°C	16.2	2.9	3.8	3.4

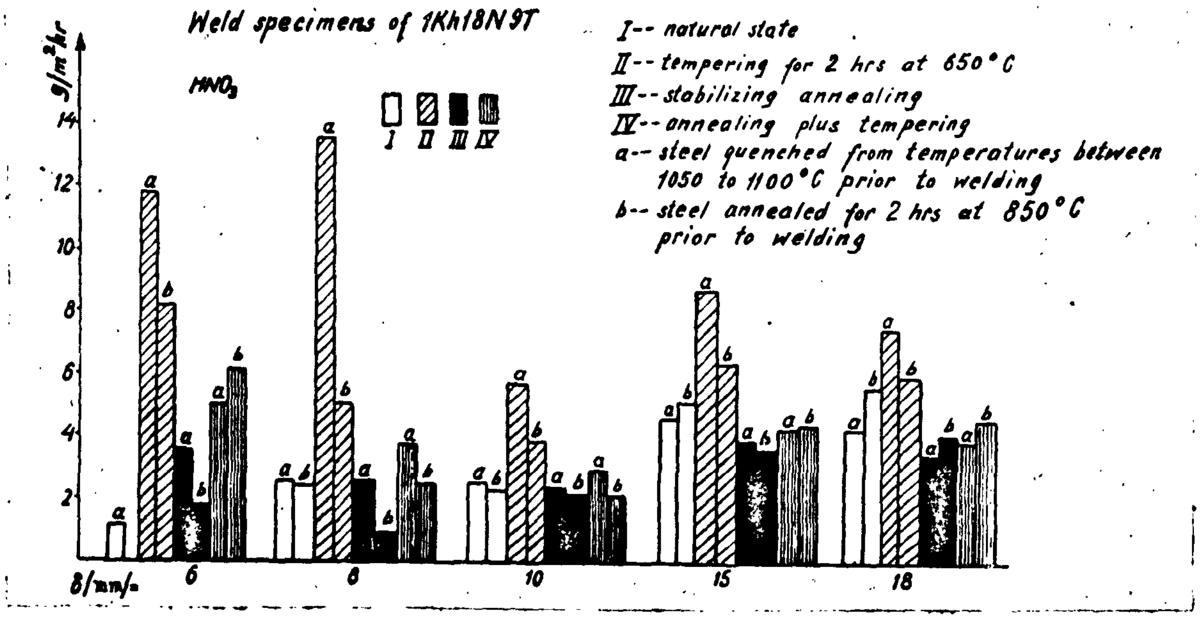

Fig. 108. The effect of stabilizing annealing on weld joints of 1Kh18N9T steel, 6 to 18 mm thick, in boiling 65% nitric acid.

Fig. 108 shows the results of tests with 5 series of 6 to 18 mm thick weld specimens of 1Kh18N9T steel after different heat treatment. The steel used in welding was subjected to quenching and stabilizing annealing. The specimens were tested in 65% nitric acid.

Tests showed that stabilizing annealing markedly increases the resistance of weld joints not only in nitric acid but in sulfuric-acid media as well. The beneficial effect of stabilizing annealing becomes most evident when the weld specimens are subsequently tempered. Here, once again, we see the invaluable advantages of diffusion (immunization) heat treatment of stainless steels, as compared with other methods of heat treatment. It should be noted that stabilizing annealing is rather effective also in regard to the chemical resistance of rolled and forged steel, thus doubtlessly increasing the chemical resistance of weld joints.

N. Yu. Pal'chuk's data concerning the effect of stabilizing annealing on the resistance of manually welded joints are in complete agreement with the above.

Another most valuable property of stabilizing annealing is its prevention of knife-line corrosion (see Section 8).

It is known that 18-8 steels are frequently used in structures doing service under deep-freeze conditions. In view of this, we tested 1Kh18N9T steel specimens preliminarily exposed to deep-freeze conditions (holding for 3 hours in liquid oxygen (-188°C)) for corrosion in boiling 55 per cent HNO_3 and in 2 per cent H_2SO_4. The results of these tests are included in Table 55.

TABLE 55. Effect of Deep-Freeze Treatment on the Resistance to Over-all Corrosion of Weld Specimens of 1Kh18N9T Steel

Heat treatment prior to corrosion tests	Loss of weight after 100-hr. boiling in g/m^2hr.	
	in 55% HNO_3	in 2% H_2SO_4
none	2.01	2.63
holding for 2 hrs. at 650°C...............	3.46	2.99
holding for 3 hrs. at -183°C	3.90	2.47
holding for 3 hrs. at -183°C, then for 2 hrs. at 650°C	5.72	3.42

These tests revealed that the treatment by cold increases the corrosion rate of 1Kh18N9T steel welds in nitric acid, and practically hardly affects the resistance of welds in sulfuric acid.

Effect of alloying elements. The effect of the alloying elements on the resistance to over-all corrosion of weld joints is not limited to their influence on the primary structure. The effect of alloying admixtures on the resistance of *welds* is, generally, similar to their effect on the resistance of austenitic rolled and forged steels in various aggressive media.

It is well know that carbon has the effect of abruptly increasing overall corrosion of stainless steels in various media. Thus, for example, Heeley and Little /179/ supply most intersting data on the effect of carbon on the resistance to overall corrosion of titanium-bearing 18-8 steel in nitric acid. The corrosion rate of steel in 30 per cent nitric acid increased tenfold with an increase in carbon content from 0.08 to 0.12 per cent (after 3 to 4 min heating at 1300°C followed by cooling off in air).

Shirley and Truman showed that increase in the carbon content of titanium- (or niobium) bearing 18-8 steels has, practically, no effect on their resistance to overall corrosion in solutions of (o.5 to 2.5 per cent) hydrochloric aid and (30 per cent) acetic acid, while in (10 per cent) sulfuric acid as well as in (70 per cent) nitric acid, corrosion rate increases sharply.

Buke, Hager, Phillips and Quineau* established that the corrosion rate of 18-8 ELC steel in boiling 65 per cent HNO_3 increases hundredfold if after holding for 2 hours at 650°C, its carbon content is increased by a mere 0.009 per cent (from 0.028 to 0.037 per cent).

Increase in carbon content, has the same effect on welds.

Hence, it is desirable to minimize the carbon content of welds both in regard to their chemical stability and their resistance to intergranular corrosion. Here, even a hundredth of a per cent is relevant. Yet carbon-content reduction, in itself, does not solve all problems. Important, and at times decisive, is also heat treatment of steels and welds. For example, Ebling and Sheil /167/ have published data according to which prolonged holding of weld specimens of ultralow-carbon steel, type 18-8, brand AISI-304 L (0.028% C, 18.06% Cr, 9.30% Ni), in 65 per cent nitric acid at temperatures between 900 and 1000°F (485 to 535°C), results in a hundredfold increase in their corrosion rate. In working conditions where they run the risk even of a short exposure to liquid media (for instance, when as a result of suspended operation, condensate may form on welds) such steels should not be used.

Of primary significance for the resistance to overall corrosion of weld joints is the chromium content of the weld metal. It should be kept in view, however, that not chromium alone but also nickel increases the resistance to overall corrosion of steels. The n/8 mole rule applies in the same measure to Fe-Ni alloys as to Fe-Cr alloys. Hence, to be resistant to overall corrosion, steels should contain 1/8, 2/8 etc. moles of nickel (13.5; 27% Ni) just as they are required to contain 1/8, 2/8 etc. moles of chromium (12.5; 25% Cr) /136/. This is the reason why the nickel content of austenitic steels designed to perform in sulfuric and phosphoric acids is increased up to 30 per cent /257/.

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* ASTM, Spec. Techn. Publ., No 93, 1950.

Since economy in nickel is desired, we tested Kh23N23M3D3 steel for welding with an electrode whose nickel content (4 to 5%) was in part replaced with manganese (6 to 7%). It proved that in the mentioned limits, this substitution had the effect of reducing somewhat the resistance of welds to overall corrosion in boiling 35 to 50, and 75 per cent sulfuric acid. Manganese impairs resistance to intergranular corrosion. Further investigation is required to discover the techniques of increasing the resistance to this type of corrosion in welds containing manganese instead of nickel.

It should be noted that manganese increases greatly the overall corrosion of welds in oxidizing media, in nitric acid inter alia (see Table 56 compiled from data of N. A. Langer and M. M. Kurtepov as given in their article "Corrosion properties of stainless-steel welds in oxidizing solutions", in "Automatic Welding", No 2, 1957).

Table 56

Effect of alloying elements on the corrosion rate of weld joints of 1Kh18N9T steel in boiling 55 per cent HNO_3

Additional alloying of weld metal	Chemical composition of weld metal in per cent						Loss of weight in g/m^2hr after 200-h boiling
	C	Si	Mn	Cr	Ni	Ti	
none	0.09	0.82	0.70	17.8	9.2	0.16	0.87
0.91% V	0.08	1.35	0.80	18.0	9.2	0.23	1.10
0.32% Nb	0.08	1.23	0.80	18.2	9.8	0.27	1.35
1.0% V - 0.32%Nb	0.07	0.56	1.37	17.9	10.1	0.27	0.93
0.98% W	0.08	0.92	0.85	18.8	10.9	0.28	6.05
1.60% W	0.09	0.78	0.99	17.6	10.2	0.29	4.88
1.61% W - 0.30%Nb	0.08	1.02	0.83	18.4	10.8	0.30	2.16
1.15%Mo - 0.28%Nb	0.09	0.61	1.38	17.9	10.1	0.30	1.09
1.80%Mo - 0.28%Nb	0.09	0.53	1.25	17.2	10.4	0.31	4.10
6.0% Mn	0.09	0.81	6.1	18.2	10.6	0.29	4.00

From the viewpoint of economizing nickel and substituting manganese for it in acid-resistant steels, the data of Flint and Toft /170/, British investigators, on the relative corrosion resistance of steels of the types 18 Cr - 8 Ni and 18 Cr - 10 Mn - 2 Ni are, doubtless, of the greatest interest. Table 57 gives the chemical composition and the mechanical properties of the said steels.

Table 57

Chemical composion and mechanical properties of austenitic steels of types 18 Cr - 8 Ni and 18 Cr - 10 Mn - 2 Ni /170/

Type of steel	Chemical composion of steel in per cent						Mechanical properties of steel		
	C	Si	Mn	Cr	Ni	Ti	σ_s in kg/mm^2	σ_b in kg/mm^2	δ in % l =50.8mm
18 Cr - 8 Ni	0.10	0.63	0.47	18.2	8.1	0.58	19.0	47.0	57.5
18 Cr - 10 Mn - 2 Ni	0.08	0.44	10.5	17.5	1.9	0.59	33.1	49.1	26.5

The results of the corrosion tests conducted in 65 per cent nitric acid and in phosphoric acid of various concentrations are presented in Table 58.

Table 58

Corrosion resistance of steels of the types 18 Cr - 8 Ni and 18 Cr - 10 Mn - 2 Ni /170/

Type of steel	Heat treatment	Corrosion rate in mg/inch^2/day in boiling:				
		65% HNO_3	10% H_3PO_4	20% H_3PO_4	30% H_3PO_4	40% H_3PO_4
18 Cr - 8 Ni	annealing	170	6	6	8	2
	tempering	546	-	-	-	-
18 Cr -10 Mn -2 Ni	annealing	566	10	17	18	27
	tempering	1368	-	-	-	-

As we see, the resistance of chromium-manganese austenitic steel in the said media is appreciably lower than that of chromium-nickel steel, while in sulfuric acid, it is only slightly inferior to titanium-bearing 18-8 steel.It proved that after having been exposed to the industrial atmosphere of Birmingham for 18 months, steel of high manganese content had rosted through and through, while 18-8 steel remained entirely intact. These authors consider that research to obtain chromium-manganese austenitic steel of a corrosion resistance not inferior to that of 18-8 steel, is necessary by all means.

The negative effect of Mn is pointed out also in work /265/.

Of late, Lula and Renshaw, American investigators, have published /191/ new data on acid-resistant chromium-manganese austenitic steels. They arrived at the conclusion that steels of the type 18 Cr - 10 Mn with merely 2 to 3 per cent of nickel (mentioned, in particular, by the British authors also) cannot compete in regard to their corrosion resistance with 18-8-type steels. It is necessary to add either 0.30 to 0.70 per cent of nitrogen, or - more effective still - to increase their nickel content up to 5 to 6 per cent. Table 59 gives the chemical composition of the latest chromium-manganese-nickel austenitic steels.

Table 59

Chemical composition of acid-resistant chromium-manganese-nickel austenitic steels /191/

Standard brands of steel	Type of steel	Chemical composition in per cent				
		C not more than	Mn	Cr	Ni	N
AISI-201	17-4-6	0.15	5.5 - 7.5	16.0-18.0	3.5 -5.5	0.25
AISI-202	18-5-8	0.15	7.5 -10.0	17.0-19.0	4.0 -6.0	0.25
-	204	0.10	7.5 -10.0	17.0-19.0	4.0 -6.0	0.25
-	204L	0.06	7.5 -10.0	17.0-19.0	4.0 -6.0	0.25
-	20-6-8*	0.10	7.5 - 9.0	19.0-21.0	5.0 -7.0	0.25 to 0.35

* Test steel

According to data /191/, an increase from 2.5 to 5% in nickel content in 17Cr - 10Mn steel brings about a triple decrease in the corrosion rate in a 65% nitric acid. Steel of Type 20Cr - 8Mn - 6Ni is only to a small degree inferior to 18-8 steel in its resistance to nitric acid as well as to acetic, sulfuric, and phosphoric acids. However, steels with a high manganese content are low-resistant to pitting and in this respect are considerably inferior to 18-8 steels. According to Lula and Renschaw, the remarkable property of chromium-manganese-nickel austenitic steels is that they may contain a higher carbon content (0.065%) than 18-8 steels and that there is no danger of tendency to intergranular corrosion after welding.

Despite the promising results obtained by the authors in their work /191/, it is necessary to state that the problem of substituting nickel by manganese in austenitic steels is not yet solved either in the USSR or in the countries abroad. Due to the necessity of using nickel sparingly and also considering the positive effect of manganese on the resistance of pure austenitic welds (not containing copper) to hot cracks, it is necessary in the very near future to find reliable means of preventing the harmful effects of manganese on the corrosion properties of austenitic steels and welds.

In connection with the development of oxygenless fluxes, at the present time in flux welding, as was mentioned earlier, it is possible to alloy welds by titanium in the same manner as it is done in gas-arc welding. It seems, however, that this is not always necessary in view of the overall corrosion resistance of weld compounds: automatic welds of 1Kh18N9T steels, which are titanium stabilized, are inferior in their resistance to boiling nitric acid than welds containing niobium.

For comparison, weld samples of 1Kh18N9T steel were subjected to corrosion tests. One group of samples was tested under oxygenless flux by a Sv-1Kh18N9T rod, the other by a Sv-1Kh18N9B rod. The chemical composition of both welds is shown in Table 60. The results of corrion tests conducted in 55% nitric acid are shown in the same table.

TABLE 60

The effect of titanium and niobium on the corrosion resistance of 1Kh18N9T steel welds in 55% nitric acid.

Type of rod	Chemical composition of the weld metal in %							Weight loss (in g/m^2 hr) after 100 hours of boiling in 55% nitric acid	
	C	Si	Mn	Cr	Ni	Ti	Nb	Without heat treatment	After tempering at 650° for 2 hours
Sv-1Kh18N9T	0.09	0.63	1.18	18.0	9.0	0.40	none	7.6	14.0
Sv-1Kh18N9B	0.09	0.64	1.00	18.0	9.6	0.28	1.02	3.2	6.5

Both welds resist intergranular corrosion, but the weld alloyed with titanium corrodes in nitric acid considerably faster than the weld with niobium. The microstructure of both welds is shown in Fig. 109. Analogous data were obtained for welds made by gas-arc welding. Thus, Benz and Sohn /156/ have indicated that welds made by the argon-arc method on 18-8 steel with niobium possess a greater resistance to over-all corrosion in boiling nitric acid than the same kind of welds made on 18-8 steel with titanium: in the natural state -- by a factor of 4, after tempering at 700°C for 2 hours -- by a factor of 9, and after hardening for austenite and tempering -- by a factor of 33. Similar data are contained in the work of Shirley and Nicholsen /208/.

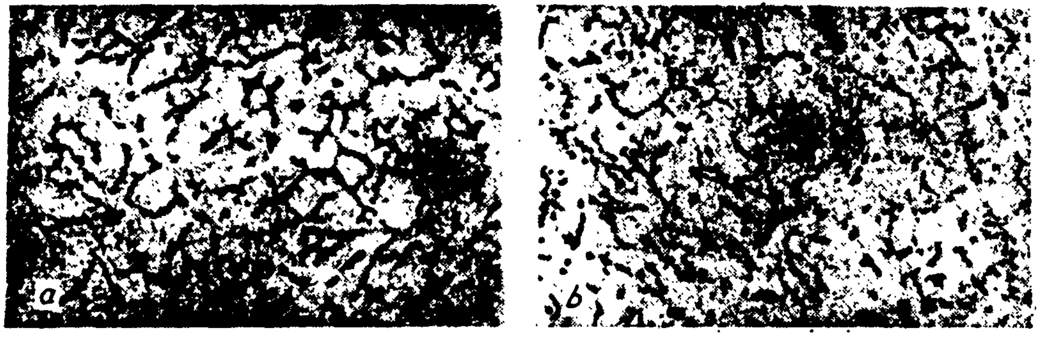

Fig. 109. Microstructure of two-phase welds of 18-8 steel which possess resistance to intergranular corrosion, but which differ in over-all resistance in nitric acid: a) weld with 0.40 titanium; b) weld with 1.02 niobium (300X).

As for the base metal, according to A. M. Samarin and A. A. Yaskevich /116/, 18-8 steel with niobium (Kh18N11B) possesses a considerably superior resistance in nitric acid as compared to 18-8 steel with titanium (1Kh18N9T).

It follows from the above that the selection of a system for alloying welds which would make welds resistant to intergranular corrosion, cannot be made without considering the effect of this system on the over-all corrosion resistance of the weld compound.

It follows from the data shown in Table 56 that in constructions which are intended for work in acidifying media, alloying welds with manganese and tungsten is not recommended. Vanadium decreases the resistance in nitric acid.

Good results are obtained from complex alloying with vanadium and niobium, as well as with niobium and molybdenum, if the concentration of the latter does not exceed 1%. In high concentrations, molybdenum has a very negative effect, decreasing the corrosion resistance of the weld (Fig. 110).

Zirconium, which prevents intergranular corrosion, unlike titanium but like niobium, somultaneously somewhat decreases the over-all corrosion of welds in 55% nitric acid. If the weld contains 0.4 to 0.8% Zr (according to calculation), the corrosion rate does not increase. However, if the weld contains 1.3% Zr (according to calculation), the corrosion rate increases by a factor of 3.

Experiments have shown that additions of small amounts of lead contribute to some increase in the overall corrosion resistance of weld compounds in boiling 65% nitric acid.

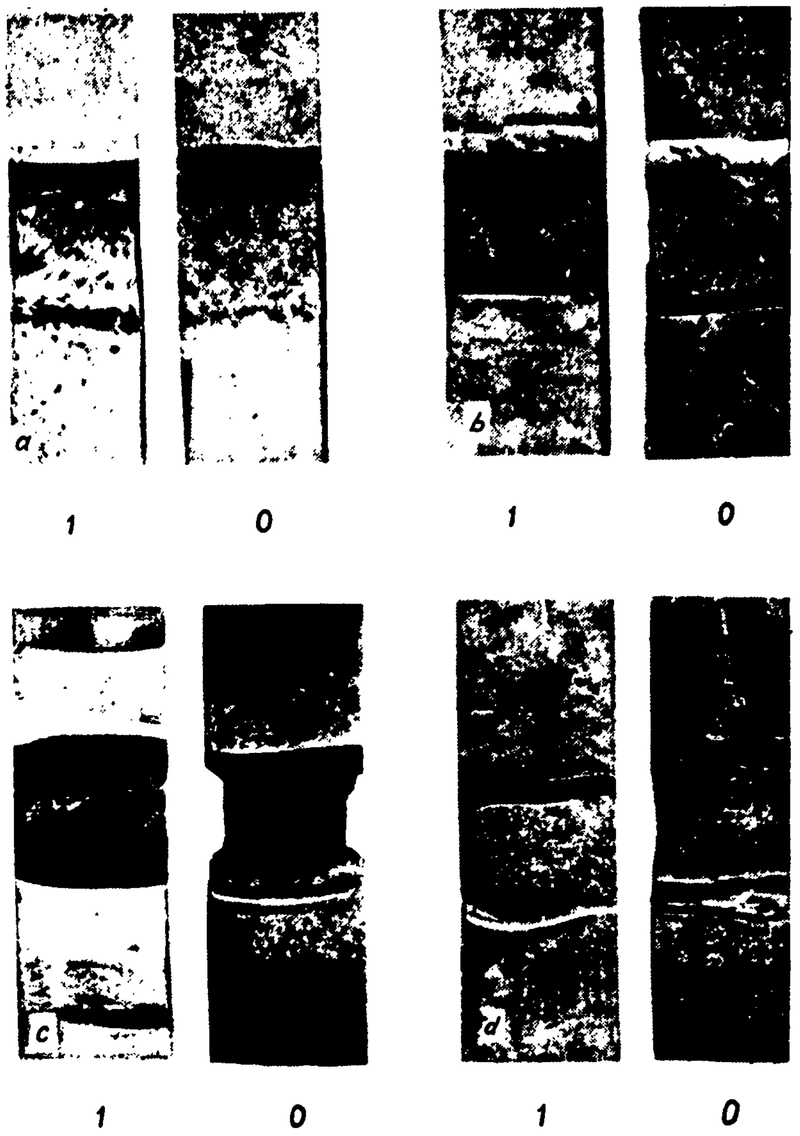

Fig. 110. The negative effect of manganese, tungsten and molybdenum on the over-all resistance of 18-8 steel welds in acidifying media:

a) weld alloyed with vanadium and niobium; b) weld alloyed with tungsten;
c) weld alloyed with molybdenum and niobium; d) weld alloyed with manganese;

1 -- natural state; 0 -- state after a two-hour tempering at 650°.

It has been noted that arsenic and sulfur very strongly decrease the overall chemical resistance of welds in sulfuric acid media.

The nature of the alloying of the weld, apparently, has a certain effect on the appearance of knife-line corrosion in the weld region near the boundaries of fusion.

There is a possibility of diffusion of certain elements from steel into the weld during the welding process, the degree of diffusion apparently depending on the kind of alloy of the weld metal.

The effect of strain hardening. The effect of strain hardening on the over-all corrosion resistance of the welds depends on the aggressive medium in which the welds operate. Together with N. A. Langer and A. G. Potap'evskiy, we found that cold tensile strain has practically no effect on the weld resistance in nitric acid, whereas in sulfuric acid media, it lowers the resistance considerably. Corrosion tests were carried out in 6% sulfuric acid solution of 10% aluminum of weld samples which had been stretched 10% (Table 61).

TABLE 61. The Effect of Strain Hardening by Stretching and of Stabilizing Annealing on the Corrosion Resistance of Welded Joints Made of 1Kh18N9T Steel After Boiling for 144 hours (Loss in Weight in g/m^{2}hour)

Aggressive medium	Heat treatment					
	No strain hardening		After strain hardening 10%			
	None	2 hours at 650°	None	2 hours at 650°	3 hours at 850°	3 hours at 850° + 2 hours at 650°
Nitric acid	0.097	0.321	0.096	0.346	0.072	0.110
Sulfuric acid solution	0.319	1.108	1.107	0.872	0.166	0.271

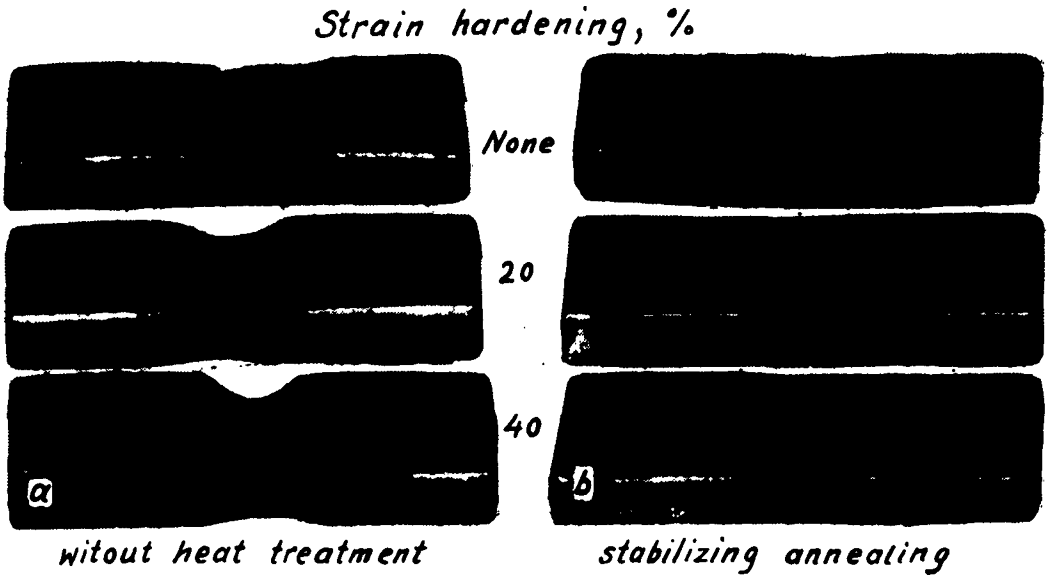

Fig. 111. Effect of strain hardening and stabilizing annealing on over-all corrosion of welds in aluminum sulfate solution in sulfuric acid.

For comparison, Table 61 gives the results of tests with specimens which, after plastic deformation, were subjected to stabilizing annealing (heating for 3 hours at 850°). It is quite clear that annealing completely neutralizes the detrimental effect which strain hardening has on the corrosion resistance of welds in a sulfuric acid medium.

Figure 111 shows the photographs of specimens stretched 40% prior to corrosion tests in sulfuric acid solution. Within 24 hours, this weld was almost completely corroded (Fig. 111 a), with a loss in weight of 32 g/m^2hour. The same weld which underwent stabilizing annealing (850° for 2 hours) after having been stretched 40%, showed hardly any corrosion (Fig. 111 b); the loss in weight was almost three times smaller (up to 11 g/m^2hour).

Analogous results were obtained in corrosion tests of specimens cut from the spherical-bottom part of an actual structure which became strain hardened as a result of cold stamping. The degree of deformation increases with the distance from the axis of the bottom to its edge: specimens from the edge of that bottom part have shown the highest degree of strain hardening. Data given in Table 62 show that strain hardening lowers the over-all corrosion resistance of the metal, and that stabilizing annealing eliminates the detrimental effect of strain hardening.

TABLE 62. The Effect of Strain Hardening (Due to Cold Stamping) and of Stabilizing Annealing on the Corrosion Resistance of 1Kh18N9T Steel in Sulfuric Acid Solution After Boiling for 200 Hours (Loss in Weight in g/m^{2}hour)

Stabilizing annealing	No. of specimens*								
	1	2	3	4	5	6	7	8	9
No	7.0	8.2	10.5	12.0	-	16.0	19.5	25.8	25.5
Yes	5.5	4.3	-	4.6	4.8	4.8	3.8	4.2	4.7

*The increase in the No. of specimens corresponds to the increase in degree of deformation.

Our conclusions on welds and austenitic steels (Fig. 112) coincide with data in the literature pertaining to the base metal. Thus, Uhlig points out that strain hardening, while influencing the corrosion resistance of austenitic steel in boiling 65% nitric acid, can bring about the intensification of corrosion in other media, e.g., in media which cause pitting corrosion. Monypenny /192/ points out that cold deformation sharply increases the corrosion of austenitic steel in sulfuric acid.

Thus, for instance, the rate of corrosion of type 18-8 steel in 25% sulfuric acid increases from 12 to 50 g/m^2hour after 20% strain hardening.

Fig. 112. Effect of strain hardening on corrosion resistance of austenitic steel in different media: a) Kh18N12M2T steel in oxidizing medium (65% nitric acid); b) Kh23823M3D3 steel in non-oxidizing medium (35% sulfuric acid).

Section 10. Transcrystalitic Corrosion in Welded Joints of Stable Austenitic Steels

In certain aggressive media, corrosion cracking of steels under the influence of tensile stresses, so-called "stress corrosion", has been observed. This type of corrosion usually has a composite character, i.e., the corrosive attack occurs both along the grain boundaries (intercystallitic corrosion) and across grains (transcrystallitic corrosion). Stress corrosion

A voluminous literature is devoted to the subject of stress corrosion in common structural steels. The problem of corrosion cracking of austenitic steels drew attention only during the recent years /108 et al/. No definite information pertaining to the stress corrosion of austenitic steel welds exists in the literature which is known to us.

G. L. Shvarts /140, 141/ has established that the austenitic 1Kh18N9T type steel, when in a state of static stress, is subject to corrosion cracking in solutions of halogen salts ($MgCl_2$, NH_4Cl, NaF, and other salts). In the presence of a concentration of stresses, the cracking sets in when the tensile stresses are near the yield strength of the steel. This steel undergoes cracking of a compound type, with a predominance, however, of intercrystalline cracks. The addition of halogen salts of oxidants $NaNO_2$ and $K_2Cr_2O_7$ to the solutions (0.01 N solutions) reduces the susceptibility of the 1Kh18N9T type steel to corrosion cracking under stress /255/.

According to literature data the transcrystalline corrosion under stress may attack chromium-nickel austenitic steels which become subjected to the effect of hydrogen sulfide (H_4S), molten Pb—Bi eutectics, as well as of water and steam.

ЕДЕЛАНИЮ

E assumes that the transcrystalline corrosion under stress in-chromium-nickel austenitic steels occurs owing to the preponderant development of small quantities of quasimartensite, which is formed in steel under the effect of plastic deformation. To reduce the susceptibility to this type of corrosion, he recommends either a) to increase the stability of the austenite by raising the nickel content of steel to 13--14% (at 0.07% C and 18% Cr), or b) to decrease its stability by lowering the nickel content to 7%. The formation of large quantities of quasimartensite in such a steel is probable. The corrosion spreads along the martensite sections on the surface of the steel, without affecting the metal in depth. Linnert is of the opinion that the two-phase ($\alpha + \gamma$) austenitic steels and welds are to the least extent susceptible to the transcrystalline corrosion under stress.

α γ

One of the most recent studies points out that corrosion cracks in the austenitic 18-8 type steel, which is in a 42% water solution of $MgCl_2$ under tensile stress almost approaching the yield strength (from 28 to 35 kg/mm^2), may develop at a rate of up to 3 mm/hr. Hoare and Hines think that one should resort to cathodic protection of the 18-8 type steel in order to prevent corrosion under stress. They deny that quasimartensite plays an important role in the spreading of corrosion cracks. Parkins /201/ holds that the best way of avoiding this type of corrosion in weld compounds (regardless of

atomic
dictionary

steel brand or type) is: a) heat treatment to relieve stresses, and b) modification of the system of stresses which act upon the metal, i. e., to replace tensile stresses by compression stresses.

The causes of metal corrosion under stress are not yet adequately investigated, which is emphasized in a recent monograph by G. L. Shvarts and M. M. Kristal'.

Nevertheless it may be ragarded as being established that the presence of tensile stresses in the parts of the matal facing the stress-producing agent is the requisite condition for transcrystalline corrosion in the austenitic steels. No transcrystalline corrosion is observed in the presence of compression stresses.

We determined, together with Yu. V. Latash and N. A. Langer, that the resistant austenitic low-alloy steels (Kh23N23M3D3, Kh23N28M3D3, Kh23N28M2T) and their welds are, in sulfuric acid, subject to transcrystalline corrosion due to the surface compression deformation.* Samples of the Kh23N23M3D3 type 3.8 mm thick steel, 20 X 70 mm large, were subjected to similar corrosion tests (Table 63), the steel had the following chemical composition: 0.06% C, 0.89% Si, 0.33% Mn, 22.25% Cr, 23.38% Ni, 2.85% Cu, 2.80% Mo, 0.010% S, and 0.010% P /77/.

One part of the samples was tested in natural state, the other part was subjected to surface cold-hardening, by pneumatic chisel with a blunt edge and by planing prior to tests. All samples which suffered compression deformation proved to be affected by transcrystalline corrosion.(Fig. 113a). One of the

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* It was determined by the X-ray analysis method that the compression deformation caused stretching stresses in the surface layers of the metal.

Table 63

Results of 100-hour boiling of cold-hardened and non-cold-hardened samples of the Kh23N23M3D3 (EI533) type steel in sulfuric acid

Sample characteristics	Thermal working before corrosion tests	Presence or absence of transcrystalline corrosion after boiling in sulfuric acid		
		35% solution	50% solution	75% solution
In natural state	None	Absent	Absent	Absent
	1 hr soaking at 1100°, water quenching	"	"	"
	2 hr soaking at 800°, air cooling	"	"	"
After cold-hardening by planing	None	Present	Present	Present
	1 hr soaking at 1100°, water quenching	Absent	Absent	Absent
	2 hr soaking at 800°, air cooling	"	"	"

samples was planed by different processes. One half of its surface was subjected to planing with a deeper penetration of the cutter (intense compression), the other half was planed by taking off only fine turnings (mild compression). After boiling in 35% solution of sulfuric acid and making a 90° bend on that part of the sample surface which was strongly cold-hardened, large corrosion cracks emerged (Fig. 113a).

Fig. 113. Transcrystalline corrosion of purely austenitic steels caused by compression deformation: a) external view of a Kh23N23M3D3 (EI533) type steel sample affected by transcrystalline corrosion (at the left is the same sample as it appeared before tests); b) structure of steel affected by the type of corrosion (300X); c) the same in a Kh23N23M3D3 type steel weld (130X).

The less strongly cold-hardened part of the sample was covered by a finer net of shallower corrosion cracks. The corrosion deterioration of cold-hardened steel (Fig. 113b) and weld metal (Fig. 113c) had a transcrystalline character.

As it should have been expected, no transcrystalline deterioration was observed during any of the tests of both kinds of samples (cold-hardened and non-cold-hardened) in $H_2SO_4+CuSO_4$ and $H_2SO_4+CuSO_4+Zn$ solutions[*], conducted for the purpose of exposing the intergranular corrosion. It should be noted that compression deformation induces in addition to the transcrystalline deterioration also a speeding up of the corrosion process of the high-alloy Kh23N23M3D3 type steels in boiling sulfuric acid (Table 64).

Experiments have shown that not only the Kh23N23M3D3 (EI533) type steel, but the similar Kh23N28M3D3 (EI629) type steel with a higher nickel content (up to 30% instead of 23%) as well, is also susceptible to transcrystalline corrosion in sulfuric acid as a result of compression deformation. The same

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[*] The addition of zinc powder to the sulfuric acid solution of copper sulfate to expose intergranular corrosion in the Kh23N23M3D3 and in other types of steel was proposed by the NIIKhIMMASh (Scientific Research Institute of Chemical Machinery).

Table 64

The effect of compression on the overall chemical stability of the Kh23N23M3D3 (EI533) type steel in boiling sulfuric acid

Sample characteristics	Heat treatment	Loss of weight after 100-hour boiling in sulfuric acid, in g/m^{2}hr		
		35% solution	50% solution	75% solution
In natural state	None	0.565	0.650	0.400
	1 hr holding at 1100°, water quenching	0.601	0.437	0.473
	2 hr holding at 800°, air cooling	0.473	0.437	--
After cold-hardening by compression	None	2.747	1,600	0.656
	1 hr holding at 1100°, water quenching	0.382	0.328	--
	2 hr holding at 800°, air cooling	0.400	0.800	--

pertains to the EI629 type steel welds. At the same time the Kh18N28M3D3 (EI530) type steel, which differs form the EI629 type steel by a lower content of chromium (18% instead of 23%), is subject to this type of corrosion deterioration to a lesser degree. The Kh23N28M2T (EI628) type steel, differing from the three types mentioned above by the absence of copper content, is also susceptible to transcrystalline corrosion after compression deformation in 35% solution of sulfuric acid.

This indicates that the presence of copper in the EI530, EI533, and EI629 type steels is not the cause of the susceptibility to transcrystalline corrosion.

It is impossible to avoid the cold-hardening and the transcrystalline corrosion susceptibility which it causes when manufacturing welded apparatus from austenitic steels. Stabilizing tempering (Table 63), which eliminates the consequences of cold-hardening of the steel and of the weld by compression, may serve as one of the ways to escape this type of corrosion. Strictly speaking, it would be sufficient to use only high-temperature tempering, i. e., heating up to 650-700°C, to eliminate the stresses which cause corrosion cracking. However, because of the danger of appearance of intergranular corrosion, it is necessary to increase the temperature of heating to 800-900°C (according to data of Yu. I. Kamennov, to 950° /36/).

II. Resistance of Welded Joints of Austenitic Steels to Intergranular and to Over-all Gas Corrosion

A voluminous literature is devoted to the problems of scale resistance. The theoretical bases of oxidation of metals are set forth in the works of V. I. Arkharov /6/ and of others /46/. Various gas corrosion test methods, providing for tests not only of non-loaded samples, but of samples which are under effect of

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dynamic loads, are developed in /139/. The tests are conducted in oxidizing and in reducing media. In recent times the conduct of experiments in those media, in which the alloy is intended to operate, is preferred. In connection with the development of gas turbine technology great attention is being paid to the problems of gas corrosion in fuel combustion products, and particularly in media containing vanadium pentoxide, which is encountered in many kinds of liquid fuel. The literature on the problem of scale resistance of welded joints is unfortunately extremely scarce. Of the studies known to us, the article by Pilarzyk, Staub, and Sniegon should be mentioned./198/. These researchers conducted gas corrosion tests of welded samples of the 25-20 type steel, placing them for a period of 90 hours into a furnace through which oxygen, at 1100°C, was blown. It proved that, in addition to the over-all gas corrosion, i. e., the scaling, [redacted] the intergranular corrosion (oxidation of the grain boundaries) also takes place. The experiments have shown that the weld is superior to the base metal in resistance to oxidation of the grain boundaries.

The scale resistance of the austenitic steel welded joints depends primarily on the chemical composition of the weld. Among the number of elements which are used for weld metal alloying in welding of austenitic steels only vanadium causes a sharp drop in scale resistance. Other metals (tungsten, manganese, molybdenum) have, according to our data, practically no effect on the scale resistance of the 25-20 type steel welds. Chromium, silicon, and aluminum boost the scale resistance of the welds.

The negative effect of vanadium can be judged from the following results. Pieces of experimental welding wire made of Kh25N15G70 steel, prepared for us on request by the "Electrostal'" plant (0.21% C, 0.21% Si, 7.37% Mn, 26.20% Cr, 16.85% Ni, 0.014% S, 0.006% P and 2.18% V), were heated at 1100°C in air. After twenty hours of heating the wire was covered with a thick coat of scale (Fig. 114a), and after 100 hours at 1100°C it had completely turned into scale (Fig. 114a). Butt welds were made under flux in Kh23N18 (EI417) steel 6 mm thick, containing various amounts of vanadium (Table 65).

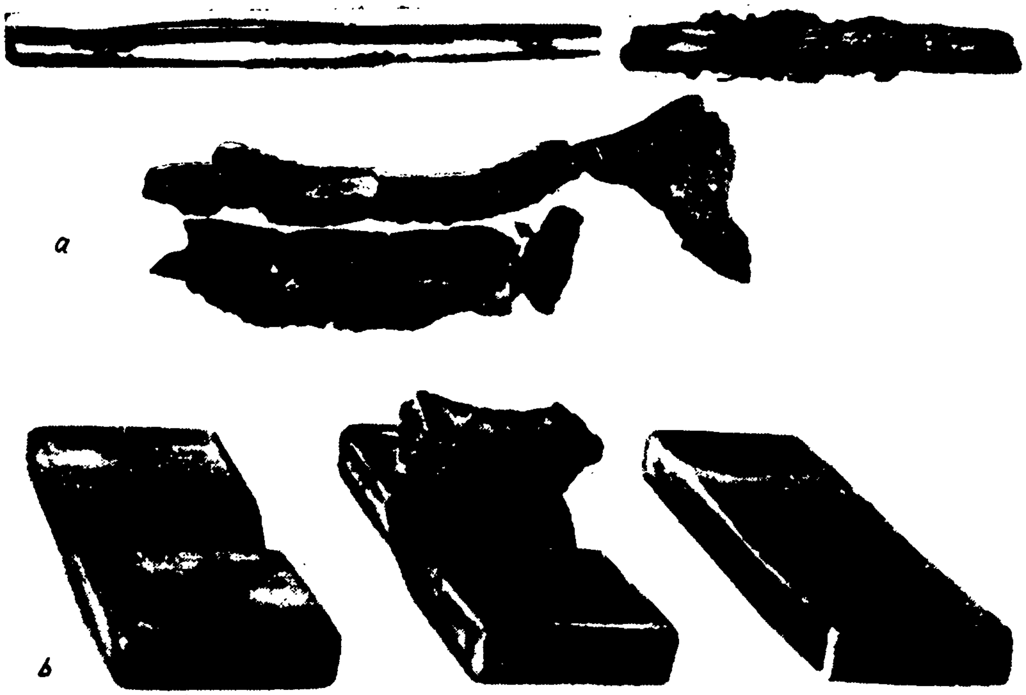

Fig. 114. Effect of vanadium on scaling when heated and held at 1100°C: a) Kh25N15G7F austenitic wire; b) 25-20 steel welds alloyed with vanadium; from left to right: before heating; after 5 hours at 1100°C; the same with scale removed.

Chemical Composition of Kh23N18 (EI417) Steel Welds Further Alloyed with Vanadium

No. of weld	Chemical composition of weld in %					
	C	Si	Mn	Cr	Ni	V
1	0.13	0.50	3.24	25.3	18.3	0.67
2	0.13	0.74	1.65	22.4	18.2	1.41
3	0.14	0.76	1.70	22.0	16.8	3.50

The welded samples were heated at 1100°C in air. After 7 hours holding noticeable scale appeared on the welds in the samples of the first series (0.67% V). After 300 hours of heating the scale-formation rate on the weld became stable, but oxidation of the weld was still more rapid than that of the steel. In the samples of the second series (1.41% V), scale 3 to 4 mm in height appeared after 100 hours of heating. In the case of the welds of the third series (more than 3% V), the coating of scale attained 15 mm after 5 hours of heating at 1100°C (Fig. 114 b); over this period the weld thinned out from 6 to 2.5 mm -- more than half the original size (Fig. 114b). Vanadium should not be used in large concentrations for alloying welds intended for operation at such high temperatures (above 900°C)*. At lower temperatures (600 to 750°) vadanium, as is known, is used in small concentrations to alloy heat-resistant austenitic steel welds. Scientific literature on the subject of scale-resistance in high-alloy steels and alloys also points out the adverse effect of vanadium. Sykes and Shirley, for example, quote data to the effect that the presence of vanadium pentoxide (V_2O_5) in a gaseous fluid causes very strong gas corrosion of austenitic steels at 750°C. The loss in weight of 25-20 steel over a period of twenty hours, for example, amounted to 20 mg/cm^2.

Berry and Fontana point out that alloys containing molybdenum, tungsten and vanadium oxidize very rapidly when they come into contact with a gaseous fluid containing the oxide vapors of these elements. The vanadium oxides have a particularly strong effect. Stainless chromium steel containing 2% vanadium oxidizes at 870 to 900°C ten times more rapidly than ordinary unalloyed carbon steel.

Fitzer together with Wetternik advocate protection of austenitic steels from gas corrosion in the presence of vanadium oxides by coating their surfaces with silicon.

Drexon, Slander, Harder and Cowe tested cast samples of chromium-nickel austenitic steels for gas corrosion at 800 to 1000°C. They note that the alloy 28Cr - 9Ni was the best. When the nickel content is high, the corrosion rate in a fluid containing sulfur increases. Silicon and aluminum slow down the corrosion rate and molybdenum and vanadium step it up slightly. Thus the general opinion is that vanadium accelerates gas corrosion. As far as we are aware, /21/ is the only work which states that a content of up to 4% vanadium has no effect on the scale-formation of iron alloys.

If there are sulfur compounds present in a gaseous fluid, austenitic steels may be affected by intergranular corrosion. As is known, the cause of this kind of corrosion is penetration by a low-melting sulfide eutectic (probably Ni_3S_2), formed by the interaction of nickel and sulfur, along the grain boundaries into the metal.

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[*] According to Yu. I. Kazennov, vanadium causes point gas corrosion in 18-8 steel welds at 650 to 700°C.

Yu. V. Rumyantsev and D. M. Chizhikov /115/ point out that the temperature at which there is intensive interaction between nickel and sulfur dioxide lies between 460 and 470°C. The reaction can be basically represented by the equation

$$7Ni + 2SO_2 \rightleftarrows Ni_3S_2 + 4NiO \quad (18).$$

The maximum depth at which the reaction occurs is reached at 600 to 800°C; a further increase in temperature inhibits the process.

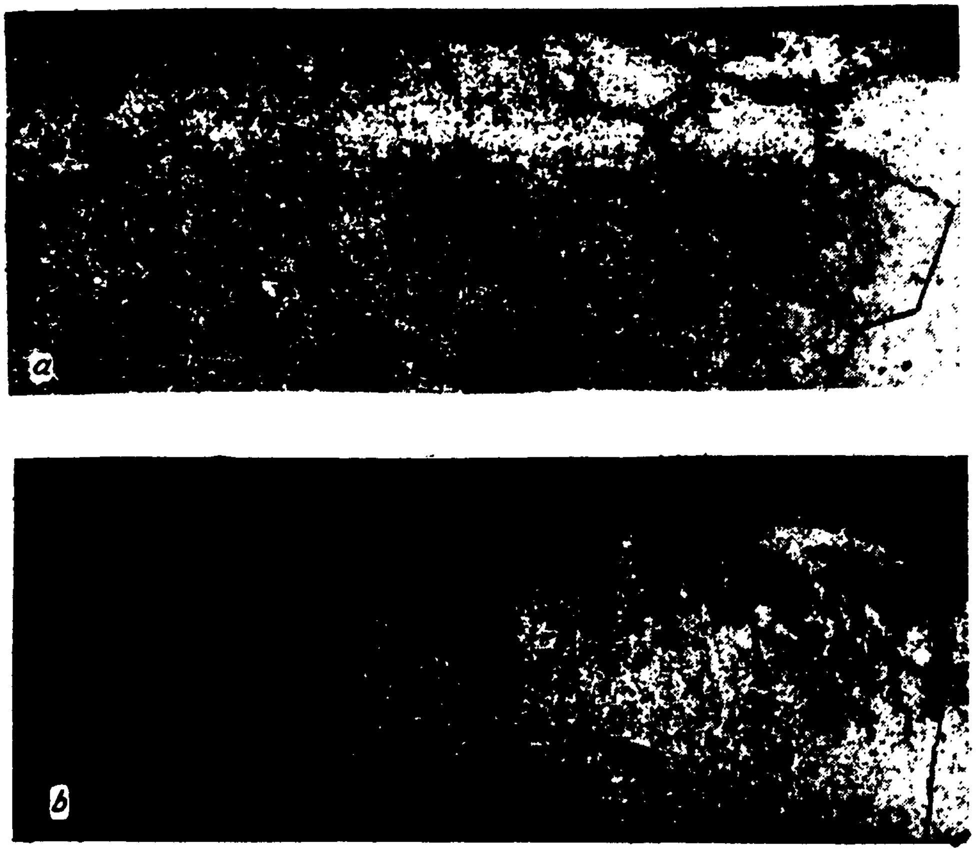

Fig. 115. Intergranular gas corrosion at 1100°C caused by sulfur:

a) Kh23N18 steel (130 x); b) Kh23N18 steel weld (130 x).

The development of intergranular sulfide corrosion is undoubtedly aided by tensile stresses and the effect of adsorption reduction of the strength of austenitic steel /112/. The more nickel there is in heat-resistant steels, alloys or welds, the greater the danger of intergranular gas corrosion occurring in an atmosphere containing sulfur compounds. Preference should therefore be given to steels, alloys and welds in which the nickel content has been reduced to minimum.

In collaboration with N. A. Langer, we tested welded samples of Kh23N18 (EI417) steel for intergranular gas corrosion by a speeded-up method. The welded samples were placed in a porcelain tube in a Silit furnace at a temperature of 1100°C. When the given temperature was reached, a porcelain boat containing a measured amount of powdered sulfur is placed in the tube with the sample. The sulfur ignites and the atmosphere becomes saturated with sulfur dioxide. After a brief period at 1100°C in this atmosphere the samples are taken out and microsections are prepared (the heating time depends on the amount of sulfur burned and the test temperature). The penetration of the sulfur dioxide into the metal can easily be seen on unetched sections (the eutectic is colored light brown). Given invariable test conditions, the resistance to intergranular gas corrosion is determined by the depth of penetration of the eutectic. In our experiments the machine welds (Fig. 115b) were not inferior in gas corrosion resistance to the base metal -- 25-20 steel (Fig. 115a). The experiments proved that partial replacement of nickel by manganese improves corrosion resistance; this is probably due to an increase in the solidification point of the sulfide eutectic.

During the test methods described the weld is subject to more difficult conditions than the steel, since it has a directional structure while in steel with equi-axial grains the eutectic spreads sideways, and not only inwards as in the steel. In order to make the conditions equal for the weld and the steel, the sulfur has to be placed in the porcelain tube at least one hour after the heating of the sample, by which time the recrystallization has taken place and the acicular crystals in the weld have given place to equi-axial grains.

On page 242 we stressed the fact that attention should be given to tests of welded joints for corrosion under working conditions. During the years 1954 and 1955 the Institute of Electric Welding, in collaboration with the Stalingrad branch of the State Design and Scientific-Research Institute for Petroleum Machinery and the Ufa Petroleum Research Institute, carried out tests for corrosion on 1Kh18N9T steel welded samples at the Novo-Ufa oil refinery under conditions involving the refinement of sulfur-containing oil. These tests, like those carried out by the Institute of Electric Welding in 1956-1957 together with the Stalin chemical combine in Dimitrovgrad (Bulgarian People's Republic) during the production of nitric acid, showed that two-phase welds alloyed with silicon, vanadium and niobium possess excellent corrosion-resistance in the fluids mentioned. The results of the laboratory tests tallied well with the tests carried out under working conditions /247/.

Chapter VI

MECHANICAL PROPERTIES OF WELDED JOINTS OF CHROMIUM-NICKEL AUSTENITIC STEELS

The mechanical properties of welds must answer different requirements, depending on working conditions. The realization of the prescribed mechanical qualities of a welded joint is achieved by proper alloying, heat treatment and welding techniques.

1. Effect of alloying, microstructure, and test temperatures on the mechanical properties of welds of austenitic steels

Welds of chromium-nickel austenitic steels, possessing a two- or multiphase structure ($\gamma + \delta$, $\gamma + K$, $\gamma + M$, and others), appreciably surpass single-phase welds in point of tensile strength, but are inferior to them in that of ductility. For this reason, the effect of alloying elements, heat treatment, plastic deformation and other factors on the mechanical properties of the weld are largely determined by the structural changes induced by these factors in the weld metal.

Mechanical properties of welds of austenitic steels at room temperature. Ferrite-forming elements, acting jointly or separately, enhance both the ultimate and the creep strength of the weld, but reduce its elongation per unit length, relative compressibility and, sometimes, its impact strength.

The effect of niobium and titanium is characterized by the data in Tables

66 and 67.

Table 66

Effect of niobium on the mechanical properties of the metal of automatic welds on 18-8 type steel

Niobium content of weld, in %	Creep strength in kg/mm^2	Ultimate strength in kg/mm^2	Elongation per unit length, %	Contraction per unit length, %	Impact strength in kgm/cm^2
None	23.9	58.7	73.5	61.4	16.7
Approx. 0.5 %..	26.5	61.4	59.0	38.6	6.9

Table 67

Effect of titanium on the mechanical properties of the metal of automatic welds on 18-8 type steel

Titanium content of weld, in %	Creep strength in kg/mm^2	Ultimate strength in kg/mm^2	Elongation per unit length, %	Contraction per unit length, %	Impact strength in kgm/cm^2
0.3	24.6	58.8	73.5	61.4	16.7
0.6	25.8	59.4	52.0	36.4	8.2

Titanium acts unfavorably on the plastic properties of the weld metal (see Table 67), but less intensively than niobium.

Silicon, aluminum, vanadium, molybdenum and other ferrite-forming agents, as distinct from ferrite-carbide forming elements (niobium and titanium, and also zirconium), enhance the tensile strength of austenitic welds of 18-8 type steel without appreciable impairment of their ductility in the natural state, that is after welding without heat treatment (see Table 68). A like action on 18-8 steel welds is produced by chromium.

It is worth noting that ferritizing admixtures, except niobium and titanium, do not have a perceptible effect on the impact strength of 18-8 type weld metal in its natural state at room temperature. However, all ferritizers ~~cause to~~ reduce impact strength at sub-zero temperatures in case the weld has a two-phase $\gamma + \delta$ structure (see below).

Chromium enhances the tensile strength, but lowers the ductility, of 25-20 type welds. Thus, e.g., with an increase in the chromium content from 25 to 30 % the elongation per unit length of the weld metal diminishes from 40 to 20 %, or by one-half.

Table 68

Mechanical properties of welds of 18-8 type steels containing ferritizers

Denomination of ferritizer introduced and its concentration in the weld, in %	Creep strength in kg/mm^2	Ultimate strength in kg/mm^2	Elongation per unit length, %	Contraction per unit length, %	Impact strength in kgm/cm^2
None.............	24.7	59.8	52.0	43.5	13.1
Silicon 1.5 % ...	28.3	65.0	48.7	41.0	-
Silicon 1.9 % ...	33.6	69.2	61.3	62.3	17.9
Vanadium 1.2 % ..	27.4	60.5	65.9	55.6	-
Vanadium 1.9 % ..	40.6	72.5	45.0	55.6	13.9
Chromium 21.5% ..	30.0	64.4	45.0	48.6	14.1
Silicon 1.2% Vanadium 1.0 Niobium 0.8	38.9	67.0	46.7	51.0	12.5

The creep and tensile strengths increase comparatively little. Fig 56 exhibits a photograph of one of Gagarin's samples cut from a weld with 32 % chromium content, which failed of brittleness during stretching. The cause of the brittleness was the formation of the -phase during the welding process.

Phosphorus adversely affects the mechanical properties of 25-20 steel welds, particularly when acid fluxes or coatings are used. Thus, for example, increasing the phosphorus content from between 0.015 and 0.018% to 0.035% and using acid electrodes, decreases the elongation per unit length of the weld metal from 40 to 20% /158, 160/. In welding with base electrodes, a comparable sharp decline in the ductility of the weld is caused by a substantially greater phosphorus content (0.050%) /160/. Silicon affects 25-20 steel welds the same way as phosphorus. In 18-8 steel welds silicon is conducive to the formation of ferrite, and consequently strengthens the weld metal. In 25-20 steel welds, silicon, as pointed out in Chapter 4, causes macroscopic and microscopic cracks, impairing the ductile and tensile strength of the welds. Phosphorus has the same effect. In order to estimate the joint effect of P and Si on 25-20 steels welds, it is recommended /160/ to use the empirical formula

$$P_e = P + 0.04\,(Si - 0.25\%)$$

where P_e is the equivalent phosphorus content, and P and Si are the actual concentrations of phosphorus and silicon in the 25-20 steel weld.

If P_e in the weld is increased from 0.025 to 0.050%, the elongation per unit decreases by a factor of 4 (from 40 to 10%). At the same time, due to the appearance of microscopic cracks, the ultimate strength is reduced by a factor of 2, i.e., from 64 to 33 kg/mm^2. In 18-8 steel welds phosphorus has a different effect than in high-nickel welds, and its adverse effect diminishes as the ferrite content in the weld increases. In 20-10 steel welds which have an austenitic-ferrite structure, an increase in phosphorus content from 0.040 to between 0.15 and 0.17% has virtually no effect on the mechanical properties according to the data of some researchers. As was shown in Chapter 4, in this case phosphorus does not cause cracks, since the presence of the -phase in the weld, due to a high relatively high chromium and nickel content, may also cause the phosphorus to have a ferritizing effect. In 18-8 steel welds hot cracks will only occur when the phosphorous content reaches about 0.6%, whereas in 25-20 steel welds hot cracks are found when it reaches only 0.045%.

Tungsten, like other ferritizers, strengthens the weld metal. Both tungsten and *vanadium* have a beneficial effect on the mechanical properties of 25-20 steel welds, whose tensile strength increases without loss of ductility (Table 69).

Molybdenum has a beneficial effect on the mechanical properties of welds of austenitic steels. Extensive use is made abroad of electrodes made of 18-8 steel with molybdenum in order to improve the weld metal ductility of both acidproof and heat-resistant steels. The KTI-5 electrodes devised by G. L. Petrov and V. V. Kyrchenov /102/ which are well-known in the USSR, as well as those recommended by Ye. V. Sokolov /127/, V. V. Ardentov and others /234/, are also intended for the alloying of welds with molybdenum on steels lacking this element. Additional alloying with molybdenum imppoves the plastic properties of the weld metal in manual arc welding of Kh18N12M3T steel: the ultimate strength increases from between 44 and 51 to between 53 and 56 kg/mm^2, the elongation per unit from between 16.5 and 40.0 to between 43.5 and 45.0, and impact strength from between 9.0 and 13.4 to between 13.2 and 14.3 kgm/cm^2 /132, 133/.

Table 69

Effect of tungsten and vanadium on the mechanical properties of weld metal of type 25-20 steels

Chemical composition of weld metal in %							Mechanical properties of weld metal			
Carbon	Silicon	Manganese	Chromium	Nickel	Tungsten	Vanadium	s in kg/mm^2	b in kg/mm^2	5 in %	in %
0.19	0.48	4.0	24.5	17.5	-	-	31.8	58.2	49.0	57.8
0.20	0.37	4.0	25.0	17.0	1.0	-	31.8	59.4	54.3	59.9
0.20	0.19	6.0	24.0	15.8	2.0	-	47.7	76.0	42.0	43.5
0.19	0.47	4.8	25.1	18.4	-	0.93	37.4	68.5	53.3	51.0

Austenitizers are helpful in improving the plastic properties of welds, unless their introduction entails hot-crack formation. For example, in a weld of type 18-8 steel, performed with a rod of brand Sv-0Kh18N9 an increase in the nickel concentration from 8 to 11 % causes a decrease in ductile strength due to the formation of microscopic cracks. On the other hand, a similar increase in nickel content with the addition of 20 to 25 % chromium does not impair the mechanical properties and may even enhance the impact strength of the weld metal.

Manganese, by reason of its ability to bind sulfur and inhibit the hot brittleness of welds, increases their ductility. This is evident from the example of manually performed welds of type 25-20 steels (Table 70). Nickel and manganese, while enhancing the ductility of welds, have practically no effect on the tensile strength indices of the weld metal.

Table 70

Effect of manganese on the plastic properties of weld metal on type 25-20 steel/160/

Manganese content in %	Per unit elongation over a length of 2", in %	Contraction per unit length in %
0.93	34.0	38.0
1.68	38.0	42.5
2.16	38.0	47.0
2.93	39.5	48.5

An increase in the manganese content in automatic welds of this type to between 6 and 8 % ensures the following tensile and ductile strength indices in these welds, viz.: $\sigma_s \geqslant 32$ kg/mm^2; $\sigma_b \geqslant 58.0$ kg/mm^2; $\delta_5 \geqslant 45$ %; $\phi \geqslant 55$ %; $a_k \geqslant 16$ kgm/cm^2.

G. A. Ukolov notes that in welds of type 18-8 steel with 2 to 4 % molybdenum manganese appreciably improves the mechanical properties of the weld metal (Table 71). The same applies also to molybdenum /133/.

Table 71

Effect of manganese on the mechanical properties of the metal of manual welds of type 18-13 steel with molybdenum /133/

With or without additional manganese alloying of weld metal	σ_s in kg/mm^2	σ_b in kg/mm^2	δ_5 in %
Without..............................	26.4	41.3	12.0—23.0
With.................................	30.4	55.2	32.0—46.0

The positive effect of manganese in the given case is apparently linked with the elimination of microscopic hot cracks due to sulfur.

Copper has virtually no influence on the mechanical properties of welds at room temperature. Some researchers hold the view that copper, being an austenitizer, weakens the σ-phase formation and embrittlement in austenitic-ferritic welds at high temperatures. However, recent investigations have proved that copper has no effect on the rate and intensity of σ-phase formation in welds of type 18-8 steels with silicon and niobium, but seems to contribute to the embrittlement of type 25-20 steel welds. Augmenting the copper content in welds of Kh23N23M3 steel renders them susceptible to superheating: the plastic properties deteriorate due to the precipitation of the cuprous excess phase along the grain boundaries. An analogous phenomenon was discovered by Yu. I. Kazennov in welds of Kh18N28M3D3B steel.

The effect of carbon on the mechanical properties of welds of austenitic steels is associated with its influence on the microstructure.

In type 25-20 welds, as was mentioned above, an increase in the carbon content, without changing the silicon content, leads to the prevention of hot cracks and thereby contributes to greater tensile and ductile strength of the weld metal. According to data contained in literature on the subject, in manual welds of type 15-35 steel an increase in carbon content from 0.05 to 0.20%, given a Si content of 0.4 to 0.6%, doubles the elongation per unit (from 18 to 38%) and increases the ultimate strength from 47 to 58 kg/mm^2. A further increase in the carbon concentration, for example, from 0.20 to 0.30%, leads to the appearance of primary carbides and increases the ultimate strength to 72 kg/mm^2 with a simultaneous decline of the elongation per unit length to between 20 and 25%. Data to the same effect are given in ref. /239/.

The beneficial effect of an increased carbon content on the mechanical properties of austenitic welds of type 25-20 steel, performed automatically under flux, is illustrated by the data in Table 72 (by comparison of welds 1 and 3). The data pertaining to welds 1 and 2 demonstrate the adverse effect of silicon.

If the carbon content in welds of 25-20 steel is raised to between 0.25 and 0.30%, the impact strength of the weld metal in as-welded state is from 20 to 26 kgm/cm^2.

Table 72

Effect of carbon and silicon on the mechanical properties of the weld metal on type 25-20 steel

Content in weld metal in %		σ_s in kg/mm^2	σ_b in kg/mm^2	δ_5 in %	ψ in %	Condition of surface of Gagarin's samples after testing
Carbon	Silicon					
0.06	0.31	33.6	51.3	21.3	26.4	Transverse tears (Fig.56)
0.07	0.73	31.8	49.0	19.0	30.6	
0.15	0.30	33.6	61.4	38.3	44.5	No tears

However, heating of such welds at 700 to 900°C even for a short time leads to a sharp fall in impact strength due to the abundant formation of secondary carbides on the grain boundaries. Thus, heating for two hours at 800°C lowers the weld impact strength from 24.9 to 7.6 kgm/cm^2.

In welds of type 18-8 steel, carbon seldom reaches a high enough concentration to cause the appearance of primary carbides. Here, the effect of carbon on the mechanical properties may become apparent in the embrittlement of the weld metal as a consequence of the precipitation of secondary carbides along the austenite grain boundaries. A weld of type 18-8 with an increased carbon content (approximately 0.17 %) was performed under conditions of decelerated cooling: the steel was preheated up to 400°C and the welding was performed with a flux backing. The action of the critical temperatures led to the precipitation in the weld of an excess phase consisting of chromium carbides and secondary ferrite lowering the plastic properties of the weld. After hardening for austenite the weld regained its ductility (Table 73).

Table 73

Structural condition of weld metal	Impact strength in kgm/cm^2 at a temperature (°C)		Remarks
	−40	+20°	
Austenite plus carbides	3.8	5.3	Natural condition after welding
Austenite	13.4	14.0	Hardening for austenite

It should be noted that the impact strength index of the weld metal in austenitic steel, determined from samples with Mesnager notches, depends, other conditions being equal, on the care taken in cleaning the notch bottom. Indeed, with thorough polishing aimed at removing the hardened metal layer, the impact strength index obtained ~~may be found to be~~ 20 to 25 % higher as compared to unpolished samples.

In this connection, it is appropriate to emphasize that the character of the microstructure under the notch has a great effect on the impact strength indices. This is indicated by test results obtained with specimens cut out from a clad-steel weld joint, consisting of low-carbon St.3 steel clad with stainless 1Kh18N9T steel (Fig. 116). In the first series of specimens the notch was situated on the side of the low-carbon steel, and in the second on the side of the stainless steel. Before the impact tests in bending, the specimens were subjected to slow heating up to 900°, ~~to~~ two hours of soaking at that temperature and subsequent slow cooling in the furnace. This heat treatment amounts to annealing in low-carbon welds; it increases the impact strength of the weld. The weld on the stainless-steel side is somewhat diluted with unalloyed steel (some melting of the St.3 steel is inevitable when welding the cladding), and intensive disintegration of austenite accompanied by a decline in impact strength therefore takes place in the weld after this heat treatment. The test results shown in Fig. 74 indicate that the tested section can exhibit sharply differing impact-strength values, depending on the microstructure of the weld metal under the notch on the specimen.

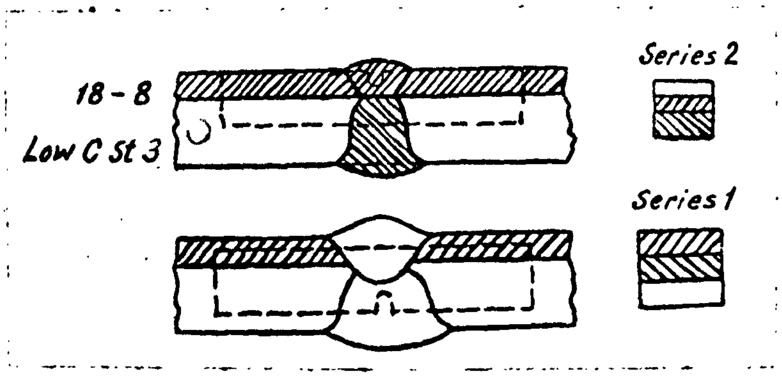

Fig. 116. Schematic drawing showing how the specimens for impact tests in bending are cut from a clad-steel weld joint.

Table 74

Effect of the weld-metal structure under the notch on the impact strength of a specimen

Series No.	Side on which the specimen was notched	Impact strength in kgm/cm^2 at	
		- 70°C	+ 15°C
1	Low-carbon steel	1.1	14.5
2	Stainless steel	2.1	7.4

Analogous data were obtained jointly by us and Yu. B. Malevskiy by comparing the results of impact tests in bending of weld specimens of 25-20-type steel which had undergone the σ-phase formation. One series of specimens were notched before prolonged heating. Here, in the sections under the notch, where the weld metal had undergone cold working from the striking edge, a greater quantity of σ-phase was formed than in the specimens in which notching was done after aging and in which the weld metal was not subjected to cold working. The specimens on which notching was carried out before aging have an impact strength 15 to 20% lower than that of specimens which were aged concurrently, but were notched after the formation of the sigma-phase.

By comparing the impact strength of the weld metal with that of 18-8 steel, we can see that the indices of the weld are always lower.

For practical purposes, austenitic and austenitic-ferritic welds in the as-welded state have equal mechanical properties, as determined by tensile tests. As far as impact strength is concerned, austenitic single-phase welds are 1.5 to 2 times stronger than welds with a two-phase austenitic-ferritic structure, but are still inferior to rolled steel.

It is known that with an increase in the cross section of single-pass welds in carbon and low-alloy steels, their impact strength declines appreciably and there is greater anisotropy of the weld metal (impact strength lengthwise and crosswise to the columnar crystals, lengthwise and crosswise to the weld /1/). Austenitic steel welds do not have this fault*.

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* According to the data of Ivao Onishi and other Japanese researchers, multi-layer welds manually welded with 18-8 steel electrodes are marked by anisotropy; they can be more or less plastic, depending on the direction in which the specimen is cut.

For purposes of comparison Table 75 gives the results of tensile tests and impact tests in bending of automatic welds made with a single pass on 1Kh18N9T steel, 10 to 50 mm thick. The chemical composition of the welds was practically equal.

Table 75

Mechanical properties of automatic single-pass 1Kh18N9T-steel welds 10 to 50 mm thick.

Thickness of steel in mm	σ_s in kg/mm^2	σ_v in kg/mm^2	δ_5 in %	ψ in %	a_k in kgm/cm^2	Bending angle of transverse sample in degrees
10 - 12	24.7	59.8	52.0	53.5	12.5 - 14.8	180
20 - 25	24.6	58.8	73.5	61.4	15.7 - 17.8	180
30	24.7	62.2	50.0	41.2	13.5 - 17.4	-
40	28.3	63.7	50.0	45.9	12.0 - 13.5	-
50	24.8	67.2	37.7	38.6	18.4 - 18.6	-

As can be seen from Table 75 that welds with a large cross-section not only are not inferior, but even surpass welds with a small cross section in impact strength. However, single-pass welds with a large cross-section, having an austenitic-ferritic structure, are to a greater extent subject to formation of the sigma-phase than welds with a small cross section and a finely-dispersed ferrite phase.

Specimens for impact testing were cut along and across the longitudinal axis of a single-pass weld with a large cross section in 1Kh18N9T steel, 50 mm thick. The test results, shown in Table 76, indicate that there is no anisotropy: the impact strength is the same lengthwise and crosswise ~~to~~ the weld. The data shown in this table for comparison indicate the magnitude of anisotropy in welds of low-carbon St.3 steel: the impact strength of specimens cut lengthwise —. . is almost half the strength across the weld.

Table 76

Impact strength of automatic single-pass welds with large cross section

Grade of steel	Type of steel	Thickness of steel in mm	Impact strength in kgm/cm^2 of specimens cut lengthwise to the weld	Impact strength in kgm/cm^2 of specimens cut crosswise to the weld
Austenitic chromium-nickel-titanium . . .	1Kh18N9T	50	17.6 - 18.9	18.4 - 18.6
Low-carbon ferritic-pearlitic	St.3	40	5.1 - 7.7	10.5 - 11.4

Multi-pass austenitic chromium-nickel steel welds have good mechanical properties. No special measures are necessary to ensure even strength of the weld, and this is true also ~~for~~ single-pass welds (it is understood that the possibility of hot cracking in the weld is excluded). Table 77 shows data ~~concerning~~ the mechanical properties of multi-pass welds of 1Kh18N9T and Kh18N11B steel, 50 mm thick. The weld was a double V butt weld.

Table 77

Mechanical properties of automatic multi-pass welds of austenitic 18-8-type steel, 50 mm thick

Type of steel	Type and diameter of electrode wire	σ_s in kg/mm^2	σ_v in kg/mm^2	δ_5 in %	ψ in %
1Kh18N9T	0Kh18N9F2S 5 mm (EI606)	40.7	71.5	56.0	55.6
Kh18N11B	Sv-0Kh18N9* 2 mm	28.3	60.1	71.0	62.3

The indices given in Table 77 pertain to specimens cut from the upper layers of the weld. If there is too much ferrite in the wire, formation of the σ-phase during the actual welding process and a decline in plasticity are possible in the lower layers; this can lead to failure of the weld when welding a rigid joint. It is therefore desirable in multi-pass welding to deposit the inner and outer beads with electrodes of different composition and different ferrite content. The strength indices of multi-layer Kh18N11B steel welds can be increased by using a

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* Twin-arc welding with separated electrodes.

OKh18N9F2S (EI606) - type wire containing vanadium.

The material given in this section indicates that austenitic steel welds made in accordance with the welding rules for these steels (see Chapter VII), regardless of whether they have an austenitic or an austenitic-ferritic structure, have high strength and plasticity indices. Changes in the welding conditions (regime and speed of welding, direction and conditions of heat transmission), provided they do not cause disintegration of austenite or cracking, usually have no detrimental effect on the mechanical properties of welds.

In some cases of a variation in the welding regime, however, and particularly when there is an inadmissibly great increase in the length of the arc in manual welding, the result can be a sharp decline in the plastic properties of the weld metal. This fact was noted when we considered the effect of nitrogen on the structure of welds (see Chapter 2). Hence, from the standpoint of the stability of the mechanical properties of the weld metal, mechanical welding which ensures a stable welding regime, is undoubtedly preferable to manual welding. Cited in literature are some very curious data on the relation between the impact strength of the metal of an austenitic weld and the skill of a manual welder. Using the same electrodes of 18-10-type steel containing 2% niobium, one welder produced a weld with an impact strength of 7.2 kgm/cm^2, while another produced a weld with a lower plasticity (3.9 kgm/cm^2). This situation cannot arise in automatic flux welding. One study quotes data concerning the effect of the electrode diameter (16-13-type steel electrode containing niobium) in manual welding on the mechanical properties of the metal deposited (Table 78).

Table 78

Effect of the diameter of a 16-13-type steel electrode containing niobium on the mechanical properties of the metal deposited.

Electrode diamter, in mm	σ_s in kg/mm^2	σ_b in kg/mm^2	δ_5 in %	ψ in %
4.0	37.5	48.2	11.4	25.0
3.25	44.8	64.8	32.5	54.0

It must be assumed that in the given case a reduction in the electrode diameter, and consequently also a decrease in energy per unit length, resulted in the elimination of hot micro-cracks caused by niobium and thus brought about an improvement in the mechanical properties of the metal deposited.

It is possible to assess the microstructure of a weld from the outer appearance of its distended surface (on Gagarin specimens or on bent specimens). An indented surface with large protruding filaments indicates a coarse crystalline austenitic weld structure. Gagarin specimens cut out from an austenitic weld are usually distended without forming a neck. High indices for necking are obtained not by a local decrease in the cross section of the specimen at the point of failure, but by a strong elongation of the entire specimen. The absence of indentation on the distended surface and the absence of necking on the Gagarin specimens is characteristic for austenitic-ferritic welds.

If there are hot micro-cracks in the weld, they can be detected on the surface of the Gagarin specimen or on the distended part of a bent specimen (Fig. 56). Specimens with multi-layer welds must be subjected to

Table 79

Effect of micro-cracks on the mechanical properties of weld metal

Type of weld steel	Microstructure of weld metal	Presence of micro-cracks in the weld	σ_s in kg/mm^2	σ_b in kg/mm^2	δ_5 in %	ψ in %
19-9	Austenite	Yes	38.0	50.0	13.0	27.8
	Austenite + ferrite	No	44.9	65.0	33.0	32.9
19-9 with niobium	Austenite	Yes	-	34.3	7.0	8.9
	Austenite + ferrite	No	43.8	65.5	40.0	50.0
25-20	Austenite	Yes	35.4	48.5	18.3	24.9
	Austenite + finely dispersed primary carbides	No	35.4	67.2	51.7	59.9

side-bending[*], in which the cracks become evident regardless of location.

If the weld metal contains hot micro-cracks, this may have an effect on its yield point and tensile strength, as well as on necking. However, the presence of micro-cracks appears most clearly in the relative elongation which is reduced from 60 - 70% to 10 - 15% or lower (Table 79).

If the hot cracks are very small, it is possible that they will not come out to the surface of the specimen, but will still affect elongation. On these grounds, some authors propose to evaluate the quality of austenitic electrodes from the results of tensile tests made with cylindrical specimens cut from the deposited metal. If the relative elongation exceeds 30%, the electrodes are considered suitable. If the relative elongation does not attain 25% they are rejected on the grounds that they will not produce a weld without hot cracks.

Intergranular corrosion has a detrimental effect on the mechanical properties of welds: strength and plasticity decline. This phenomenon was put to use by Japanese researchers /197/ in evaluating the corrosion resistance of welds.

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[*] Translator's note: Bending of a thin specimen, taken transversely across the welded joint, so that the length of the weld is perpendicular to the plane of the specimen (Metals Handbook, p. 139)

Specimens for tensile tests and impact tests in bending, were cut from a weld and boiled in a solution of copper sulfate and sulfuric acid. The chemical composition of the weld metal was: 0.08% C, 0.65% Si, 2.05% Mn, 18.15% Cr, 12.10% Ni, 2.43% Mo and 1.45% Cu.

The weld joint in its natural state had the following properties: σ_b = 60 kg/mm^2; δ_5 = 50%; a_k = 20 kgm/cm^2. The strength and plasticity of the specimens which were attacked by intercrystallitic corrosion after boiling considerably declined: σ_b = 15 kg/mm^2; δ_5 = 8%; a_k = 6 kgm/cm^2.

Mechanical properties of austenitic steel welds at negative temperatures.

The high mechanical properties of austenitic low-carbon steel welds are pre-determined by the composition of these steels, i. e., by high alloying and by the absence of secondary crystallization (under normal welding conditions). Therefore, to ensure the required properties of the weld metal is a considerably less complex task than to produce welds free from hot cracks and resistant to liquid and gas corrosion. This pertains mainly to welds doing service at room or subcritical temperatures (up to 350°C). The situation is different with welds that are used under conditions of extreme cold or at high temperatures.

Welds doing service at negative temperatures must have a high impact strength. It is due to the very fact that austenitic steels preserve a surprisingly high impact strength at extremely low temperatures (Table 80) that they are used under conditions of deep freeze. However, as mentioned above, welds, particularly austenitic-ferritic ones, are inferior to steel in this property even at room temperature. With respect to all other mechanical properties, welds are not at all inferior to steel under conditions of deep freeze (Table 80).

Table 80

Mechanical properties of 1Kh18N9T steel and austenitic-ferritic welds at at + 20 and - 196°C

Material	Testing temperature, in °C		Mechanical properties			
		σ_s, in kg/mm^2	σ_b in kg/mm^2	δ_5 in %	ψ in %	a_k in kgm cm^2
Steel	+ 20	31.0	54.7	58.3	65.0	28.5
Weld		25.9	59.0	57.0	50.9	13.1
Steel	- 196	42.4	136.1	34.4	19.0	23.5
Weld		35.4	136.0	30.0	16.0	6.9

The composition of the steel was: 0.10% C, 0.5% Si, 1.0% Mn, 17.5% Cr, 9.4% Ni and 0.57% Ti; the chemical composition of the weld was: 0.10% C, 0.53% Si, 0.47% Mn, 17.5% Cr, 8.5% Ni and 0.12% Ti.

As can be seen from Table 80, the impact strength of the steel hardly changed during the temperature drop from +20 to -196°C. At the same time, the impact strength of the weld was almost halved.

The impact strength of welds at room temperature can be raised to 18-20 kg/cm^2 by austenitising them by means of heating and hardening at temperatures ranging from 1100 to 1200°C. Owing to the stable austenitic structure, the high impact strengths imparted by hardening are retained even at low temperatures.

Table 81

Effect of weld structure and test temperatures on impact strength

Steel type	Chemical composition of weld metal in %									Weld structure	Impact strength in kg/cm^2 at	
	C	Si	Mn	Cr	Ni	Ti	Nb	Mo	Cu		+20°C	-196°C
1Kh18N9T	0.08	0.57	0.44	17.6	10.8	0.16	-	-	-	A+F[*]	12.3	4.7
Kh18N12M3T	0.07	1.44	0.78	19.3	11.4	0.29	0.83	1.50	-	A+F	8.1	3.2
Kh23N18	0.15	0.22	1.5	25.5	18.9	-	-	-	-	A	18.2	16.0
Kh23N23M3D3	0.07	1.03	0.41	24.5	23.3	-	-	3.06	3.16	A	20.6	17.0

The impact strength indices of various steel welds having in the natural state after welding single-phase and two-phase structures are adduced for comparison in Table 81.

Kh23N18- and Kh23N23M3D3-type stable austenitic steel welds have a high impact strength at negative temperatures. According to Thomas[**] this also applies to 308L- and 304L-type steel welds.

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[*] A = austenite, F = ferrite

[**] Metal Progress, July, 1956.

T. M. Slutskaya, Yu. N. Gotal'skiy and I. V. Novikov /120/ have shown that welds effected with Ti-bearing welding rods /(1Kh18N9T, Kh20N10G6T (EI613)/ become unserviceable at very low temperatures. Their impact strength is far too low, and even hardening could not appreciably raise it. At -183°C the impact strength is 2.0 (without heat-treatment) and 6.7 kg/cm^2 (after hardening at 1100°C). We believe that in this particular case the cause for the ineffectiveness of hardening has-to be sought in the high δ-ferrite content of the weld. In Chapter 2 we pointed out that in order to austenitise such welds they had to be held for a considerably longer time at temperatures raised up to 1200°C. Reference /120/ also notes that single-phase austenitic welds effected with Kh20N15- and Kh20N14G6-type steel rods have an impact strength of 30.7 and 33.8 kg/cm^2, respectively, at +20°C, and 19.7 and 23.4 kg/cm^2, respectively, at -183°C.

For a weld containing 18% Cr and 8% Ni, an impact strength value better than 6 to 7 kg/cm^2 at -183 or -196° (boiling points of liquid O and N, respectively) is not obtainable without hardening. Hence for the welding of 18-8 steels in service at very low temperatures rods with higher content in austenitizers (Ni and Mn) will have to be used. According to S. K. Zvegintsev*** excellent results in manual welding of

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***Private communication

1Kh18N9T-type steels are obtained with Kh16N26M26 (EI395)-type steel rods. Our experiments show that Kh20N80T (EI437)-type steel rods can be successfully used for the same purpose. In the case of automatic square butt welding, the above-mentioned 1Kh18N9T steel rods provide an impact strength of at least 8 kgm/cm^2 at -196°C. In v-butt welding, i.e., with large quantities of welding metal in the weld, the impact strength of the latter can be increased, but hot cracks are likely to occur.

Mechanical properties of austenitic steel welds at high temperatures. At high temperatures, austenitic steel welds possess quite satisfactory mechanical properties. Of course, the continuous influence of temperatures ranging from 350 to 875° while bringing about structural transformations in the weld metal may affect its tensile and ductile strength. This, however, will be discussed in more detail in the next chapter. In Table 82 we set forth the results of short-time tensile tests on Gagarin's specimens cut from the weld. Welds of 1Kh18N9T steel made with EI606 rods differ from those made with Sv-0Kh18N9 rods by containing roughly 1.0% vanadium. In welding Kh18N11B steel with an EI606 rod, the weld contains 0.92% V and 0.92% Nb when there is 19.2% chromium.

Table 82

Effect of test temperatures on mechanical properties of short duration of 18-8 steel welds

Brand of steel	Brand of rod	Test temperatures in °C	σ_s in kg/mm²	σ_b in kg/mm²	δ_5 in %	ψ in %	a_k in kg/cm²
1Kh18N9T	Sv-OKh18N9	+20	26.5	57.6	60.0	59.9	13.2
1Kh18N9T	OKh18N9F2C	+20	35.4	64.6	52.2	59.0	12.4
Kh18N11B	(EI606)	+20	40.7	69.0	46.2	52.1	12.5
1Kh18N9T	Sv-OKh18N9	500	14.1	41.0	43.3	59.9	-
1Kh18N9T	OKh18N9F2C	500	24.0	49.5	36.0	36.0	-
Kh18N11B	(EI606)	500	26.5	54.5	33.3	30.6	-
1Kh18N9T	Sv-OKh18N9	650	16.0	37.5	33.8	48.6	14.5
1Kh18N9T	OKh18N9F2C	650	21.2	48.4	32.4	51.0	12.8
Kh18N11B	(EI606)	650	24.7	50.5	31.4	48.6	14.9
1Kh18N9T	Sv-OKh18N9	750	12.4	20.2	28.5	48.6	-
1Kh18N9T	OKh18N9F2C	750	16.6	31.8	24.1	30.6	-
Kh18N11B	(EI606)	750	21.2	34.6	18.5	34.9	-
1Kh18N9T	Sv-OKh18N9T	850	10.6	13.0	19.7	49.9	-
1Kh18N9T	OKh18N9F2C	850	14.1	20.5	11.2	22.0	-
Kh18N11B	(EI606)	850	18.4	25.2	11.2	16.0	-

These data indicate that at high temperatures niobium and vanadium considerably increase the short-enduring strength of welds without any appreciable impairment of their ductility. Whereas from the standpoint of the mechanical properties of welds at room temperature niobium cannot be considered a desirable addition to the weld metal (see Table 66), it is nevertheless quite essential in welds designed for service at high temperatures. Niobium increases the long-enduring strength of welds at high temperatures (Table 83). For this reason, modern power plants operating at high temperatures, tend to introduce niobium into welds in spite of the fact that it may cause hot cracks and intensify the $\delta \rightarrow \sigma$ transformation.

Table 83

Effect of niobium on long-enduring strength of weld metal at 650° (stress 24 kg/mm^2). /221/.

Nb content of 19-9 weld metal in %	None	None	Nb = 5C*	Nb = 10C	Nb = 15C	Nb = 20C
Weld microstructure	Austenite	Austenite + ferrite	Austenite + ferrite	Austenite + ferrite	Austenite + ferrite	Austenite + ferrite
Time until failure in hours	5-8	10-15	400	300	600	520

*C. Nb is the C and Nb content of weld.

It is noteworthy that at 650° the long-enduring strength of austenitic-ferrite welds (without niobium) is somewhat longer-enduring than of purely austenitic welds. It should be noted in this connection that according to data of several investigations, at 600°C the resistance to intergranular failure of 18-8 two-phase austenitic-ferrite steels under stress is higher than that of 18-8 austenitic steels. After 100 hours service at 600°C the relative elongation per unit of two-phase steel dropped from 50 to 45%, and that of single-phase steel from 25 to 15%. After 150 hours service at 600°C the relative elongation rate of single-phase steel dropped from 12 to 5%, but remained unchanged at 17% in two-phase steel.

Molybdenum, tungsten and vanadium act similarly to niobium and improve the long-enduring strength of welds at high temperatures.

The best result of strengthening austenitic weld metals is attained by combined alloying with several elements such as vanadium, niobium and molybdenum, or molybdenum, tungsten and vanadium, etc. This strengthening is attained not merely on account of the appearance of the ferrite component, but also as a result of the increase of the strength of the austenitic root of the weld. Strengthening of the weld metal, attained by multicomponent alloying, becomes even stronger, the higher the temperature at which it is worked.

I. I. Kornilov conclusively proved on samples of solid nickel solutions the effective influence of combined alloying on the heat resistance of welds (Table 84). Whereas a tri-component alloying makes it possible to increase the strength of the alloy at 700 to 900° by a factor of 7 to 10, an increase in the number of components from 3 to 5 gave as high a strengthening of the solid solution as by a factor of 12 to 40 /38/. Welds, containing up to 5% ferrite, regardless of the $\delta \rightarrow \sigma$ transformation, give an excellent performance under stress at temperatures between 600 and 650°C. Their tensile strength and creep limit is also quite satisfactory. With a higher ferrite

Table 84

Effect of the number of alloying components on the tensile strength of solid Nickel solutions at high temperatures /36/

Material	Limit of tensile strength in kg/mm^2 and strengthening rate at temperatures in $^{\circ}$C					
	20	400	600	700	800	900
Pure Ni	38	22	12	7	3	1.5
Two-component Ni + 20% Cr system	81.9	76	63.0	47.0	28.0	15.0
Strengthening rate vs. pure Ni	2.17	3.5	5.25	6.7	9.3	10.0
Three-component Ni+20%Cr+3%Ti system	93	95	79	69	35	20
Strengthening rate vs. pure Ni	2.45	4.30	6.6	9.8	11.7	13.3
Five-component Ni+Cr+Ti+Al+W system	100	102	90	87	85	60
Strenghening rate vs. pure Ni	2.63	4.6	7.5	12.4	28.3	40.0

content the embrittlement caused by the formation of the σ-phase becomes critical. For this reason both in the USSR and abroad the use of electrodes warranting a strictly limited ferrite content in heat-resistant austenitic steel welds is presently recommended. According to K. B. Lyubavskiy, B. I. Lazerev and others /58, 241/, TsT-7 brand electrodes of this type in which the ferrite content can be controlled satisfy all requirements for the mechanical properties of welds on Kh14N14V2M (EI257) austenitic steel. The approximate composition of the weld metal is: 0.10% C, 0.2% Si, 2.5% Mn, 18% Cr, 12% Ni, 2% Mo, 1% V; with an optimum ratio between the chromium and nickel contents ranging from 1.45 to 1.6, and the ferrite content in the weld varying between 1.5 to 2 and 4.5 to 5%.

According to /55, 58/ the long-duration tensile strength of welds effected with electrodes of brand TsT-7 is not inferior to that of rolled 1Kh14N14V2M (EI257)-type steel at 600°C (no less than 16 kg/mm^2 after 100,000 hours of conventional testing) or to the creep limit at the same temperature (no less than 10 kg/mm^2 at creep rates of 1.10^{-5}% per hour).

Wylie, Corey and Leyda /229/, who investigated the long-duration tensile strength of various austenitic steel welds at 600 to 650°C, ascertained that in this respect the welds of 19-9 austenitic-ferritic steel with niobium are superior to the welds of 25-20 austenitic steel (without niobium). In the former, heating for 100 hours at 900°C caused the long-duration tensile strength to drop owing to the $\delta \rightarrow \sigma$ transformation (Table 86). In these welds the σ-phase forms mainly inside the grains. In austenitic welds, by contrast, the σ-phase precipitates on the intergranular boundaries after heating for 100 hours at 790°, exerting a stronger effect and noticeably reducing the long long-duration tensile strength of the weld. It is significant that at 650°C the two-phase welds with niobium can endure 1.5 times greater stresses for periods three times longer than single-phase welds of 25-20 steel.

The author, together with A. N. Safonnikov and R. O. Lents /249/ has tested the heat resistance of austenitic steel welds made by flux welding in argon and carbon dioxide, as well by electroslag welding (Table 85). It was found that with the same raw materials (steel, wire), the greatest stress-rupture strength at high temperatures is obtained by welding in carbon dioxide, because of carburization of the weld metal (by 0.02 to 0.04%) and hardening of grain boundaries by carburizing.

TABLE 85. Stress-Rupture Strength of the Welds of Chromium-Nickel-Titanium Steel 1Kh18N9T

Welding method	Type of welding wire	Chemical composition of weld metal in %			Test conditions		Passage of time before rupture in hours
		C	Ti	Nb	temperature in °C	stress in kg/mm^2	
Flux welding in argon and CO_2	Sv-OKh18N9	0.11	0.20	-	600	30	1.5
		0.10	0.24	-	600	30	3
		0.12	0.7	-	600	30	4
Flux welding in argon and CO_2	Sv-1Kh18N9T	0.11	0.32	-	600	30	35
		0.11	0.45	-	600	80	21
		0.14	0.30	-	600	30	218
Flux welding in argon and CO_2	Sv-1Kh18N9B	0.10	0.21	0.63	600	30	33
		0.11	0.23	0.80	600	30	86
		0.13	0.18	0.48	600	30	More than 320
Electroslag welding	1Khk89T (plate-like electrode)	0.07	0.40	-	700	18	7
		0.10	0.44		700	18	42

If there are no strong carbide-formers present in the weld, the effect of the increase in heat resistance is small. The increase in the carbon content of the weld during flux welding in argon, or electroslag welding, (up to 0.13 to 0.15%), also increases its heat resistance if the weld contains titanium, niobium, tungsten, vanadium, molybdenum. A much higher

TABLE 86. The Effect of Isothermal Heating (Aging) on the Stress-Rupture Strength of the Weld Metal at 650°C /229/

Type of weld metal	Aging conditions	Stress in kg/mm^2	Passage of time prior to rupture in hours
19-9 with niobium	Without aging	20	693
	100 hours at 650°	20	783
	100 hours at 900°	20	231.5
25-20	Without aging	14	237
	100 hours at 650°	14	341
	100 hours at 790°	14	123

heat resistance in welds is obtained with oxygen-free fluoride fluxes or by welding in argon than with silicate welding fluxes.

In the opinion of the authors of study /229/, electrodes which would afford, at 650°C or more, the same stress-rupture strength in welds, as in base metal (rolled or wrought steel), have not been constructed as yet.

Meanwhile, such electrodes have been developed in the USSR /103, 239, and others/.

Section 2. Effect of Heat Treatment on the Mechanical Properties of Welds of Austenitic Steels

Heat treatment effects the mechanical properties inasfar as it causes certain structural transformations. We have seen that hardening of welds which in as-welded condition contain carbides, results in an increase in the plasticity of the welds. Hardening of austenitic-ferritic welds has little effect on their properties at room temperature; they are sufficiently high even in as-welded condition.

However, heating to the region of critical temperatures (350 to 875°C) causes a sharp decrease in the plasticity of the weld metal in many cases.

The embrittlement of the weld metal can result from aging under the influence of temperatures within the range of 350 to 550°C (so-called "475-degree brittleness"), precipitation of secondary carbides at the boundaries of crystals or crystallites of the austenite, or as a result of the formation of σ-phase.

Isothermal heating at 330 to 550°C causes a sharp decrease in the impact strngth of welds containing more than 20 to 25% σ-phase, with a concentration of chromium of no less than 20%, and when alloyed with aluminum, titanium, niobium, vanadium, and silicon. In purely austenitic welds containing 0.04% carbon, the "475-degree brittleness" does not occur, though, according to data in the literature, extra-low-carbon austenitic steels of 18-8 type become brittle after heating at 450 to 550°C.

TABLE 87. Effect of Prolonged Heating at 400 to 475°C on the Impact Strength of Austenitic-Ferritic 18-8 Steel Welds (kgm/cm^2)

	Heating temperature in °C									
	Without heating	400		450				800 hrs. at 450° and 1 hr. at 950°; quenched in water	475	
Duration of heating in hours	-	24	24	84	272	500	800		18	42
Impact strength in kgm/cm^2 at 20°C after heat treatment	11.9	6.3	2.9	2.5	1.3	1.9	1.4	10.0	3.5	5.0

Table 87 gives data pertaining to the influence of heating at 400 to 475° on a two-phase weld. The composition of the weld metal is: 0.009% C, 2.1%Si, 1.5% Mn, 20.2% Cr, 8.0% Ni, 1.47% V, 0.54% Nb.

The impact strength lost by the weld as a result of heating at 350 to 550°C can be recovered by subsequent hardening at low temperature (900°C). If the weld contains titanium, the temperature of the hardening must be increased to 950 to 1000°C to avoid an intensification of brittleness which can result from the $\delta \rightarrow \sigma$ transformation.

Prolonged heating at 650 to 875°C of purely austenitic welds of type 25-20 steel results in a considerable lowering of the impact strength due to $\gamma \rightarrow \sigma$ transformation. It must be pointed out that σ-phase, while causing a sharp drop in impact strength, has comparatively little effect on the characteristics of the weld metal, which are determined by the static tension (see Table 93).

Data show that with as little as 2 to 3% of σ-phase present in the metal, the impact strength drops sharply. An increase in the amount of σ-phase up to 20 to 30% has a less intensive effect.

The formation of σ-phase in single-phase welds of type 25-20 steel is strongest at temperatures of 800 to 850°C.

At 700°C the $\gamma \to \sigma$ process is less intensive. According to data in literature, in type 25-20 steel less than 1% [illegible] phase forms even after 5000 hours of heating at 1200°F (650°C). Table 88 shows the results of impact tests on bending of automatic welds of [illegible] 25-20 steel which were subjected to prolonged heating at 700 and 800°C. The chemical composition of the welds: 0.14% C, 0.36% Si, 1.79% Mn, 23.2% Cr, 20.0% Ni. Some of the specimens were subjected to strain hardening by 30% stretching prior to heating.

TABLE 88. Effect of Prolonged Heating at 700 to 800°C on the Impact Strength of Austenitic 25-20 Steel Welds

Strain hardening	Impact strength (in kgm/cm²) at + 20°C after heating (in hrs.) at temperatures:												
	Without heating	700°C			800°C								
		60	200	1100	24	48	72	100	200	300	400	500	850	3000
No	18.4	11.1	11.4	10.9	14.7	14.7	14.5	14.5	11.4	8.2	7.2	5.9	6.1	5.5
Yes	7.2	9.2	-	4.9	7.6	9.0	8.4	6.5	5.5	3.9	4.0	3.7	2.0	2.1

¶ We have indicated above (see Chapter II), that manganese and tungsten intensify the embrittlement of 25-20 steel welds (Table 89).

It must be noted that heating at 800 to 875°C, while lowering the impact strength at room temperature, has a considerably smaller effect on a_k at the heating temperature indicated (Table 90).

Data in Table 90 show also that subsequent heating at high temperatures (at 1000°C) recovers the high impact strength of the weld metal, which it had lost as a result of aging at 800 to 850°C.

TABLE 89. Effect of Manganese and Tungsten on the Impact Strength of Austenitic 25-20 Steel Welds After Prolonged Heating at 800°C

Type of weld metal	Strain hardening 30%	Impact strength (in kgm/cm°C) at + 20°C after heating (in hrs.) at 800°C						
		None	100	200	300	500	1000	3000
25-20	No	18.4	16.7	13.2	14.5	10.4	8.1	5.5
	Yes	6.7	7.5	7.3	5.0	3.0	3.0	2.3
25-20 with 4% Mn and 1.9% W	No	18.7	12.9	8.6	5.5	3.3	2.4	1.1
	Yes	5.7	3.3	1.7	1.5	1.0	1.0	0.8

TABLE 90. Effect of Prolonged Heating at 800 to 850°C on the Impact Strength at Different Temperatures of Type 25-20 Steel and Austenitic Weld Metal

Material	Aging prior to testing (3000 hours at 800 to 850°C)	Hardening after aging	Impact strength (in kgm/cm^2) at temperature in °C	
			+ 20	850
Type 25-20 steel	No	No	18.3	9.8
	Yes	No	4.0	8.1
	Yes	Yes*	14.0	-
Weld metal	No	No	15.8	14.4
	Yes	No	2.3	8.8
	Yes	Yes**	17.3	-

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*Heating prior to hardening 2 hours at 1000°

**Heating prior to hardening 4 hours at 1000°.

Heating at temperatures ranging from 550 to 875°C does not exhibit any marked effect on mechanical properties of 18-8 steel welds with single-phase austenitic structure or a ferrite content of up to 5%. Nevertheless, two-phase welds with higher ferrite content are very strongly embrittled. The impact strength decreases more sharply, the higher the heating temperature. From the data in Table 91 it is apparent that a 2-hour soaking of the weld alloyed with niobium, silicon, and vanadium (0.08% C, 2.1% Si, 0.89% Mn, 19.8%Cr, 10.1% Ni, 0.82% Nb, 1.36% V), at 750 and 850° causes almost as sharp

TABLE 91. Effect of the Soaking Temperature on the Impact Strength of Austenitic-Ferritic Welds of Type 18-8 Steel

Impact strength (in kgm/cm^2) after heating at temperature (in °C) and soaking (in hrs)				
No	650;2	650;200	750;2	850;2
8.6	7.2	1.3	2.5	2.5

a drop in impact strength as a 200-hour soaking at 650°C. The weld in as-welded condition contained more than 25% δ-phase.

In Chapter III we showed thata 2-hour heating at 850 to 875°, so-called stabilizing annealing, has a very positive effect on the corrosion resistance of welds. At the same time, it can bring about very intensive embrittlement of the welds due to $\delta \rightarrow \sigma$ transformation (Table 92). Therefore, if the weld contains a large amount of ferrite, or if highly alloyed with titanium, aluminum, zirconium, niobium, vanadium, the temperature of the stabilizing annealing must be somewhat increased (up to 925 to 950°C). Table 92 gives data on the effect of the annealing temperature on the impact strength of the weld, confirming what was said before.

TABLE 92. Effect of Annealing Temperature on the Impact Strength of Two-Phase Welds of Type 18-8 Steel

No. of welds	Chemical composition of the weld metal in %						Impact strength (in kgm/cm^2) after 2-hour annealing at temperature (in °C)					
	C	Si	Mn	Cr	Ni	V	Without annealing	850	875	900	925	950
1	0.07	2.3	1.18	19.1	8.9	none	16.0	11.6	14.7	16.1	18.0	15.7
2	0.08	2.2	1.19	21.0	8.0	2.3	11.6	3.0	4.8	11.4	15.4	11.1

The effect of prolonged (up to 4000 hours) soaking at 650° on the mechanical properties of 18-8 steel welds containing different amounts of ferrite in as-welded condition is illustrated by data in Table 93.

TABLE 93. Effect of Prolonged Heating at 650°C on the Mechanical Properties of Austenitic-Ferritic 18-8 Steel Welds

No. of welds according to Table 28 (p. 111)	Type of weld metal	Amount of δ-phase in the weld in as-welded condition in %	Soaking at 650°C in hrs.	σ_s in kg/mm²	σ_b in kg/mm²	δ_5 in%	ψ %	a_k in kgm/cm² At temperature + 20	a_k in kgm/cm² At temperature 650
3	18-8 with 1.28% Si	Up to 10	0	27.4	64.0	69.7	57.7	17.9	16.7
			500	31.8	71.0	53.9	59.9	10.9	-
			4000	28.3	70.0	46.7	41.2	7.5	12.6
4	18-8 with 0.95% Nb	3 to 5	0	27.6	58.0	61.7	44.7	10.4	15.4
			500	33.6	63.6	52.2	53.3	8.9	12.3
			4000	30.0	65.5	41.7	30.6	4.8	10.5
1	18-8 with 0.45% Si	3 to 5	0	24.8	59.8	52.0	43.5	13.1	-
			500	26.5	62.0	59.7	61.2	9.9	-
			4000	23.0	61.2	50.0	64.0	9.8	-

Prolonged heating at 600° causes some hardening of the welds and decrease in the relative elongation and contraction, and has an especially unfavorable effect on impact strength at room and subzero temperatures. At the soaking temperature, the welds retain a fairly high degree of impact strength. G. L. Petrov and V. V. Kyrchenov /102/ note that austenite-ferritic welds, the impact strength of which decreased to 4 kgm/cm^2 at room temperature after being heated for 1000 hours at 600°, have fairly acceptable characteristics at 600°. Similar data are contained in /58/ and others.

Hence, the decrease in the weld-metal impact strength, which is caused by the transformation $\delta \rightarrow \sigma$ or $\gamma \rightarrow \sigma$, is of greatest danger to the welded structures not at the operating temperatures, but during the period of temperature decrease, for example, if the unit is shut-down.

It is regrettable that one cannot say the same of welds which become brittle as a result of aging at 350 to 500°C. Thus, for example, a weld with a two-phase structure which had an initial impact strength of 8.6 kgm/cm^2 becomes very brittle after 188 hours of work at 475°C not only at the room temperature (0.3 kgm/cm^2), but also at the operating temperature (0.9 kgm/cm^2).

It was pointed out in Chapter 3 that the degree of sigma phase formation in two-phase welds is determined not only (or not so much) by the amount of ferrite, but also by its quality characteristics (the system of alloying, extent of alloying with certain elements, dispersion of the σ-phase). Given equal chromium contents and equal amounts of σ-phase, in the initial state the welds which were alloyed with titanium, aluminum, zirconium, niobium, vanadium, molybdenum are most liable to embrittlement because of the $\delta \rightarrow \sigma$ transformation. Silicon has a considerably weaker effect on 18-8 steel welds than the above elements.

The experimental data in Table 94 gives clear indication of the unfavorable effect of niobium.

TABLE 94.

Effect of niobium on the impact strength of two-phase austenite-ferritic welds of 18-8 steel after aging at 700 and 800°C.

Weld No.	Chemical composition of the weld metal in %						Impact strength (in kgm/cm^2) after aging (in hours) at temperature (in °C)						
	Nb	C	Si	Mn	Cr	Ni	Without aging	700					800
								65	450	725	2000	3500	850
1	0.41	0.09	0.48	1.58	19.2	11.0	14.4	10.9	8.6	6.9	4.8	3.9	5.9
2	0.64	0.10	0.47	1.95	18.8	11.8	13.0	11.6	5.6	4.4	1.8	--	3.9
3	0.88	0.09	0.56	1.90	19.6	11.0	14.4	5.4	3.5	2.9	1.8	1.4	1.8

The data obtained by Kauhausen and shown in Table 95 give clear indication of the unfavorable effect of molybdenum on the impact strength of weld metal which works at high temperatures. The same table gives the results of impact tests in bending on welded samples made with 16-13 austenitic steel electrodes. One of the welds (No. 1) does not contain molybdenum, while the other (No. 2) has 2.01%.

TABLE 95.

The effect of molybdenum on the impact strength of 16-13 steel welds after aging at 650°C (Kauhausen's data).

Weld No.	Chemical composition of electrodes used in welding of No. 1 and No. 2 welds in %							Impact strength (in kgm/cm^2) at +20°C after aging (in hours) at 650°C				
	Mo	C	Si	Mn	Cr	Ni	Nb	Without aging	200	600	800	1400
1	None	0.10	0.37	2.19	15.8	12.5	1.14	10.4	9.0	7.5	7.0	7.2
2	2.01	0.11	0.33	2.34	16.3	13.2	1.11	9.0	3.0	2.0	2.0	1.0

The weld without molybdenum retained acceptable impact strength even after 1400 hours of aging at 650°C. The weld alloyed with molybdenum became brittle after 200 hours of heating at 650°C.

Welds made with Sv-Kh25N13 or Kh25N13 electrodes containing niobium become more brittle than welds made with 18-8 steel electrodes. This is furthered by the increased amount of ferrite in the weld metal. One should refrain from using the above electrodes (Kh25N13 and Kh25N13B) in welding of heat resistant steels.*

TABLE 96

The effect of single-pass weld cross section (dispersion of δ-ferrite) on its impact strength after the $\delta \rightarrow \sigma$ transformation.

Type of steel	Type of wire	Steel thickness in mm	Impact strength at 20°C in kgm/cm^2	
			In natural state (without aging)	After 32 hours of heating at 850°C
1Kh18N9T	Kh20N10G6T	12	14.0	6.3
	(EI613)	50	11.0	2.5

The data in Table 96 give an indication of the effect of the dispersion of δ-ferrite, and, consequently, of the σ-phase on the impact strength. Here we can compare single-pass welds of approximately the same chemical composition, but made under different conditions.

A weld with a small cross section in 12 mm thick steel, where the σ-phase is more dispersed, becomes brittle to a lesser degree than a weld

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* Yu. I. Kazennov (on the basis of experiments with welds containing 0.12% C, 0.77% Mn, 0.88% Si, 23% Cr, 11.8% Ni, 0.97% Nb), and also G. A. Ukolov and V. M. Pushkarev came to the same conclusion.

with a larger cross section in 50 mm thick steel, where the σ-phase exists in the form of large sections. Because of the dispersion of the ferrite component the welds are usually less sensitive to prolonged heating than cast austenitic steel of the same chemical composition.

Preliminary austenitizing of two-phase welds in 18-8 steel (hardening at 1050 to 1200°C) decreases the amount of δ-phase, or wholly eliminated it, and makes the welds more immune or completely immune to the unfavorable effect of critical temperatures. Relevant data concerning automatic welds in Kh18N11B steel are given in Table 97. The chemical composition of the weld metal is: 0.08% C, 1.67% Si, 1.0% Mn, 19.2% Cr, 9.7% Ni, 1.01% Nb, 0.66% V.

TABLE 97

The effect of preliminary hardening on the impact strenth of two-phase welds of 18-8 steel subjected to prolonged heating at 650°C.

Preliminary hardening from 1100°	Impact strength (in kgm/cm^2) after a period (in hours) of heating at 650°C		
	Without heating	200	1000
None	11.6	3.6	2.5
Yes	14.9	11.9	9.0

The hardening that follows at 900 to 1000°C, as it was pointed out in Chapter 3, restores the mechanical properties of welds, which have changed as a result of the sigma-phase formation (Table 98).

TABLE 98

The effect of subsequent hardening from 900 to 1400° on the impact strength of welds after aging for 2235 hours at 650°

Chemical composition of the weld metal in %						Impact strength (in kgm/cm^2) after aging and subsequent hardening				
C	Si	Mn	Cr	Ni	V	None	After 1 hour of heating at 900°	After 1 hour of heating at 1000°	After 1 hour of heating at 1100°	After 10 minutes of heating at 1400°
0.08	1.7	0.81	18.8	8.1	2.3	2.5	13.9	15.0	13.9	17.8

It should be noted in conclusion that at the present time there is no one opinion on the effect of the σ-phase on the serviceability of welds at high temperatures. There are data which show that over comparatively short periods of service of the metal before the transformation $\delta \rightarrow \sigma$ or $\gamma \rightarrow \sigma$ is complete, the σ-phase does not lower the heat resistance if the σ-phase in dispered and not concentrated along the grain boundaries. Later on, the presence of this phase can have an unfavorable effect on the heat resistance on account of coagulation of the σ-phase and softening of the metal. In any case the appearance of the σ-phase is admissible in the weld metal of 18-8 steel containing 5% of σ-phase working at temperatures up to 650°C /243, 252/.

3. Effect of Cold Hardening on Mechanical Properties of Chromium-Nickel Austenitic Steel Welds

Austenitic steels differ from all other types of structural steel in that they can be cold hardened to an exceptional degree. It is widely known that this property of austenite is used to advantage in parts made from high-manganese Hatfield steel (12 to 13% Mn). As a result of plastic deformation chromium-nickel austenitic steels and welds become stronger and less plastic. Cold hardening greatly improves the hardness of austenitic welds. The increase in hardness may be fairly high no matter whether it occurs as a result of a martensitic or ferritic transformation.

Machine-arc welds in 1Kh18N9T steel were strain hardened by 10 to 40% stretching /61/. The results of subsequent tensile and bending-impact tests on samples cut from the cold hardened welds are given in Table 99.

The effect of strain hardening on the mechanical properties of the weld metal of 18-8 steel.

Strain hardening	σ_B in kg/mm^2	σ_b in kg/mm^2	δ_5 in %	ψ in %	a_k in kgm/cm^2	Hardness N_B
None	32.5	60.5	60.0	55.6	$\frac{10.2 - 11.7}{10.9}$	149
10	36.7	62.0	54.7	64.0	$\frac{7.8 - 8.2}{8.0}$	207
20	50.8	69.9	54.7	55.6	$\frac{4.7 - 5.5}{5.1}$	241
30	62.1	76.2	43.5	55.6	$\frac{2.8 - 4.0}{3.4}$	255
40	70.6	79.0	28.0	55.6	$\frac{2.5 - 2.9}{2.7}$	262

The chemical composition of the weld was 0.11% C, 0.55% Si, 0.94% Mn, 17.1% cr and 10.8% Ni.

It follows from these data that the yield stress of the weld metal was more than doubled as a result of 40% strain hardening, the relative elongation contracting to the same extent.

It was pointed out above (see Chapter 2) that compressive strain causes a more intensive decrease in the plasticity of the weld metal than tensile strain. Thus, for example, in cold stamping approximately the same amount of deformation (40%) causes an increase in the yield stress from 33.2 to 89.0 kg/mm^2, i.e., 2.5 times, and a decrease in the relative elongation by almost six times -- from 57.7 to 10.7%, and not by half, as in stretching. In strain hardening by stretching the relative contraction of the weld metal does not vary for practical purposes, whereas in strain hardening by compression there is a reduction of the relative contraction by almost half.

The negative effect of hardening on the plastic properties of the austenitic-steel weld metal may result in a brittle fracturing of welded structures during bending, punching and pressing. This impedes considerably the production of austenitic-steel fittings and joints.

Superior mechanical properties of austenitic-steel welds, lost through har-

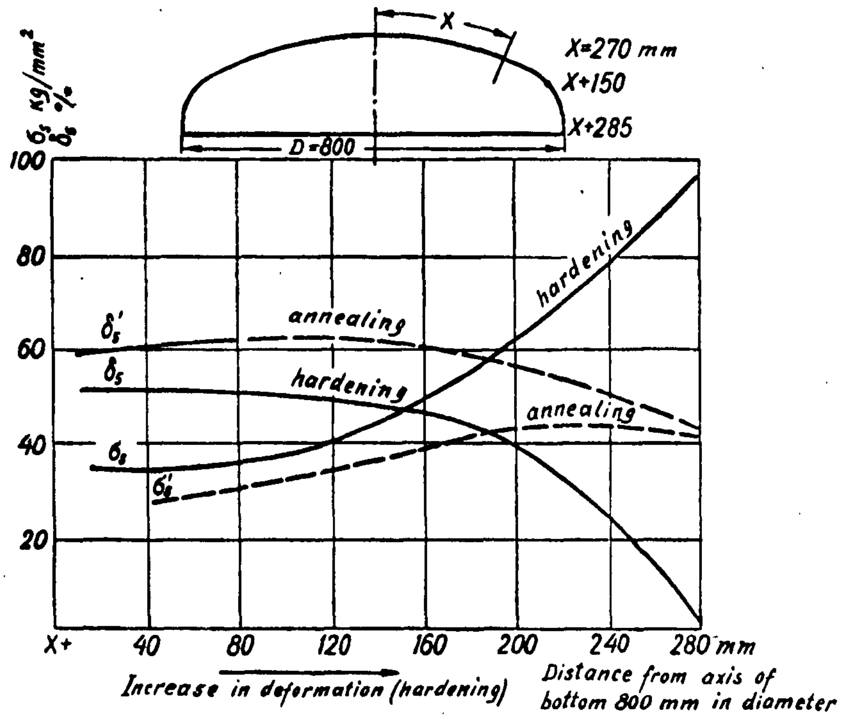

Fig. 117. Effect of hardening in cold stamping and subsequent stabilizing annealing on mechanical properties of 18-8 steel weld metal:
solid lines - no thermal treatment after hardening; dotted line - 2-hour stabilizing annealing at 850° and air-cooling.

dening, may be recovered by way of quenching for austenite from 1100-1150°C. As it was indicated, hardening is, however, undesirable in view of such serious operational disadvantages as intensive scaling, a warping of the structure and the occurrence in it of additional stresses, due to the practically unavoidable nonuniformity of the heating and air-cooling of the object. Our investigations showed that hardening was not necessary; what suffices is a mere stabilizing annealing (see Chapter III), i. e., heating up to 800-900° and air-cooling. Heating of hardened welds to 800-900° causes an intensive mellowing of ferrite. Heating to such temperatures causes a coagulation of the excess-phase particles precipitated at the grain boundaries during hardening. Moreover, fine grains formed during hardening assume a more uniform, rounded shape, and the internal stresses deactivate. As a result, the weld recovers its superior mechanical properties. Fig. 117 presents curves illustrating the beneficial effect of stabilizing annealing (two-hour treatment at 850°C) on the plasticity and strength of a 18-8 steel weld hardened by cold stamping.

To prevent the σ-phase formation in welds whose welded structure contains the δ-phase in excess, the temperature of stabilizing annealing must be raised to between 900 and 950°C.

The adverse effect of hardening on the plastic properties of the weld metal can eliminated by two-hour annealing or by tempering at a temperature lower than 850°, for example, at 650° to 750°C. This can, however, be attained only when the δ-phase content in the weld is not over 7 to 10%. More prolonged heating at these temperatures may cause further deterioration of the plastic properties, this time due to the formation of the σ-phase. The welds, whose composition is given in Table 30, were 10, 20 and 30% strain-hardened by stretching and then differently heat-treated. The structural transformations caused by the hardening and subsequent heating are described in Chapter 3, Section 5. The results of impact tests in bending are presented in Table 100.

Weld No. 0 is not subject to the σ-phase formation. The impact strength lowered by hardening is restored after two-hour heating at 750 to 850°C.

Prolonged heating at 700 to 850°C, despite the hardening, is not only safe but even beneficial to the ferrite-poor weld No. 0.

Hardening is dangerous for sigmatization-susceptible welds. This is evident from the example of 25-20 steel welds (Table 88). It is also dangerous for weld No. 9, relatively rich in ferrite which was alloyed with niobium and vanadium; two-hour tempering at 650 to 850°, including at 700°C restores partially the impact strength of the hardened weld; but after only 75 hrs heating at 700°C the weld again becomes

Table 100

Effect of hardening and subsequent heat treatment on the shock viscosity of two-phase austenitic-ferrite 18-8 steel welds.

Weld No	Hardening %	Shock viscosity (kg/cm²) at 20°C after heat treatment									
		No heat treatment	2-hour tempering at °C				Isotermic heating at 700°C for (hrs)				
			650	700	750	850	2	75	220	600	1100
0	None	12.6	15.7	16.4	14.1	14.7	16.4	11.9	11.1	13.2	14.1
	10	9.5	10.4	13.6	13.0	14.0	13.6	10.6	10.0	12.5	15.4
	20	6.3	8.4	8.7	10.0	11.9	8.7	9.3	10.0	12.9	15.0
	30	4.1	5.6	6.3	8.9	13.1	6.3	7.1	8.0	12.4	15.0
9	None	13.3	11.9	14.1	11.9	14.4	14.1	11.4	10.9	6.0	6.5
	10	8.0	9.3	9.8	9.8	11.9	9.8	8.6	5.0	3.5	3.1
	20	4.8	7.1	6.5	6.5	8.4	6.5	5.4	4.1	2.9	2.0
	30	3.8	5.6	5.0	5.6	7.8	5.0	2.5	2.2	1.9	1.9

brittle, due to the formation of the σ-phase. On hardening, the weld 9 is found to be magnetized to a rather considerable degree, its shock viscosity being 3.8 kg/cm^2. On 75-hour heating at 700°C, the shock viscosity of the weld is still less (2.5 kg/cm^2), and the weld is nearly amagnetic which is an evidence of the occurrence of $\delta \rightarrow \sigma$ conversion.

Stamping (forging) of austenitic welds strengthens them substantially. So, according to literature data, stamping of 25-20 steel austenitic welds with a ball-die pneumatic hammer yielded the following results (Table 101):

Table 101

Effect of duration of stamping on the mechanical properties of austenitic welds [171]

Stamping duration sec	5	10	30	60	120
σ_b, kg/mm^2	65.5	67.1	68.5	68.5	74.3
δ %*	39.5	30.0	28.0	33.0	20.0

* Specimens 50.8 mm long.

As it was shown in [58], austenitic welds may be hardened spontaneously under the action of shrinkage stresses during welding. This phenomenon, observed first by F. I. Pashukanis, may occur during the welding of austenitic-steel rigid solid joints. Spontaneous hardening strengthens the weld metal and reduces its plasticity. Data on spontaneous hardening of welds obtained with TsT-7 electrodes are given in Table 102 (data of TsNIITMASh, Central Scientific Research Institute of Technology and Machinery).

Table 102

Effect of rigidity of welded joint on the mechanical properties of the metal of weld obtained with TsT-7 electrodes (data of TsNIITMASh)

Welding conditions	Mechanical properties of weld metal			
	σ_s kg/mm^2	σ_b kg/mm^2	δ_s %	ψ %
Common, without rigid fastening	38.5	60.5	40.0	50.0
Rigid joint welding	52.7	66.9	27.4	49.2

The degree of the spontaneous hardening of the weld metal corresponds in this case to a hardening from 15% stretch. It is of interest that the austenization (water-quenching on a two-hour heating at 1050°C) of rigidly fastened welds causes recrystallization, i. e., the occurrence of a fine equiaxial structure. Nonhardened welds, treated similarly, show no recrystallization. The recrystallization of 25-20 steel austenitic welds, 10, 20 and 30% hardened, is in evidence in the microphotographs in Fig. 46.

CHAPTER 7

TECHNOLOGICAL PROBLEMS OF WELDING CHROMIUM-NICKEL AUSTENITIC STEELS*

Section 1. Basic peculiarities of the technology and technique of arc welding of chromium-nickel austenitic steels.

The specific physical properties of austenitic steels, such as their lower heat conductivity, electric resistance, and high coefficient of linear expansion, all cause a higher degree of warping during welding. Therefore, the main rule applicable to the welding of austenitic steels is to use methods which possess a maximum concentration of heating. Arc welding makes it possible to obtain concentrated heating and for this reason it causes less warping than gas welding. Automatic and semiautomatic welding under flux and gas-shielded arc welding are, in this respect, superior to bare arc and gas welding.

In connection with a lower heat conductivity and high electric resistance, other conditions being equal, an austenitic electrode wire melts faster than the regular carbon rod (it has a higher coefficient of fusion). It is therefore a must in automatic welding of austenitic steels to decrease the protrusion of

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* Technological problems of welding of austenitic steels have been discussed in detail in the author's brochure, Mashgiz 1958, series "Biblioteka svarshchika" (Welder's Library).

the electrode wire by a factor of 1.5 to 2 as compared with welding of carbon steels. During welding with an austenitic wire of a 5 mm diameter and a current of 500 to 700 amp., the electrode protrusion should not exceed 50 mm. In welding with a thin wire of a diameter from 1.6 to 2.00 mm, there is a particularly acute need to reduce the protrusion (at $d_{\text{electrode}} = 2$ mm, protrusion should be 20 to 25 mm).

An increased electric resistance of austenitic steel makes it necessary to use relatively short electrodes in manual welding (200 to 350 mm long instead of the usual 450 mm, according to All-Union State Standard 2523-51), and low welding currents /195/.

The poor heat conductivity of austenitic steels makes it possible to increase the welding speed by using a vertical electrode without fear that there might appear such a dangerous defect in the formation of the weld as deep and sharp edge undercuts, the so-called zone of poor fusion /1/.

The lower heat conductivity of austenitic steels permits welding to be performed with currents weaker than those recommended for plain steels. With one and the same welding current the depth of fusion of austenitic steels is greater than that of low-carbon steels.

18-8 steels stabilized with titanium or columbium can become prone to knife-line corrosion through excessive superheating and decelerated cooling of the base metal near the welding line. For this reason it is recommended to effect multipass welding of thick acidproof steels in a manner ensuring an appropriate intermediate cooling of the product. It is desirable that the side of the v-top facing the aggressive medium be welded last.

2. Technological means for preventing hot cracks in welding chromium-nickel austenitic steels

As indicated in Chapter 4, a radical way of preventing the formation of hot cracks in chromium-nickel austenitic steel welds consists in influencing the metallurgical welding processes by an appropriate supplementary alloying of the weld metal, and in applying oxygen-free and low-silicon fluxes and basic electrode coatings. In practice, in default of electrodes of the composition required, resorting to technological means for preventing the formation of hot cracks will become necessary in many cases.

These means are:

1) changing the proportions of welding and base metal in the weld metal; 2) preheating; 3) reducing the cross section of the weld and changing its shape; 4) peening (stamping) of weldeable edges or underlying layers.

Changing the proportion of the welding and base metal can only be effective if the formation of hot cracks is caused by a transition of admixtures from the base metal into the weld. A salient example is provided by the welding of Kh18N11B-type acidproof Ni-Cr-Nb steel. It is known that niobium engenders the formation of hot cracks. In automatic flux welding of Kh18N11B-type steels, under ordinary conditions, with a standard wire of brand Sv-0Kh18N9 the niobium concentration rises to highly critical points (0.4 to 0.5%) and hot cracks occur in the weld. However, cracks may be avoided if the niobium content is reduced at least by one half. To do this the base metal portion in the weld has to be reduced by one half, viz., from γ = 0.6 to 0.7 customary to all flux weldings effected under ordinary conditions to γ = 0.3 to 0.4 (thereby reducing, of course, the stability of the weld against intergranular corrosion).

A diminution of the base metal portion in flux welding is obtainable with the aid of the following techmological procedures/1/: a) v-butt welding; b) welding with gap; c) welding with (forehand) inclined electrodes, and with inclined components;

d) multiple arc welding with separated electrodes; e) welding with direct polarity d. c. current; f) welding with increased arc voltage; g) welding with thin wire; h) welding with split electrodes.

A substantial lowering of the base metal portion can be obtained with v-butt welding. It was this fact which some time ago enabled I. Z. Kagan to effect manual welding of Kh18N11B-type steels with niobium-free electrodes. However, v-butt welding reduces the output of the process. The same drawback is also inherent in welding with gap.

Twin-arc v-butt welding with a thin electrode yields the best results: a maximum reduction of the base metal portion ($\gamma = 0.25$ to 0.35) is obtainable without impairing the output of the process.

In twin-arc v-butt welding of chromiun-nickel-niobium steels with standard electrodes of brand Sv-OKh18N9 crack-free welds are comparatively easy to obtain. It should be kept in mind, however, that in single-arc welding of Kh18N11B-type steels with the same electrodes hot cracks will occur in the weld.

Preheating is an effective means for preventing the formation of hot cracks in carbon steel welds. It is known that by preheating medium-carbon steels at temperatures up to 150-200° the formation of hot cracks in the weld will be avoided. The positive effect of preheating is usually linked to decelerated crystallization of the weld pool and the abatement of the force factor due to the deferment of the moment of intensification of the tensile stresses to the critical point. /1/. Experiment shows that the action of preheating is less effective in the welding of austenitic steels than in that of plain steels. In many cases preheating enhances cracking rather than inhibiting it. In this connection it is interesting to note that at the Welders' Conference of 1955 in Frankfort on the Main /186/ it was pointed out that preheating of Nb carrying 16-13-type austenitic steels at temperatures up to 500-550°C, still in use quite recently, has a harmful effect: indeed, it enhances the formation of hot cracks. In multipass welding of austenitic steels it is therefore advisable to lower the preheating temperature to 100-150°C. Yet, according to S. K. Zvegintsev and E. A. Sirota /32/, preheating at temperatures up to 300°C appears to be a helpful factor in welding Kh18N25S2 (EYa3S)-type austenitic carbide Ni-Cr-Si steels containing 0.30 to 0.40% C. In flux welding of 6 mm thick T-specimens of Kh23N18 (EI417)-type steels, preheating inhibited the formation of hot cracks. However, in automatic welding of Kh18N12M3T-type steels, even with preheating at temperatures up to 450°C cracking could not be avoided. Preheating up to 150°C proved useful in the welding of 25-13-type cast steels. Kh15N35-type steels were welded with the aid of preheating at temperatures up to 700°C.

There is not, and connot be, a universal recommendation as to the optimal temperature of preheating in welding austenitic steels.

The effect produced by preheating or concurrent heating may differ depending on the composition of the weld metal, properties of the steel, and other factors. It was pointed out in Chapter III that preheating decelerates the primary crystallization in the weld pool; in consequence, it is bound to influence the degree of dendrite liquation in the weld. We are of opinion that the different (useful or harmful) effect of preheating in austenitic steel welding can be explained by its influence on the amount and nature of distribution of the eutectic component in the weld.

The adverse action of preheating may also be attributable to a higher content of harmful admixtures in the weld metal owing to a greater depth of fusion in the base metal. It is for this very reason that excessive preheating in welding niobium-alloyed austenitic steels with niobium-free electrodes intensifies cracking of the weld metal.

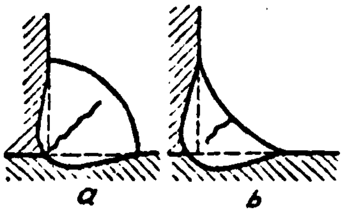

Fig. 118. Effect of shape of weld on disposition of hot cracks:
(a) in convex weld: interior cracks;
(b) in concave weld: exterior cracks

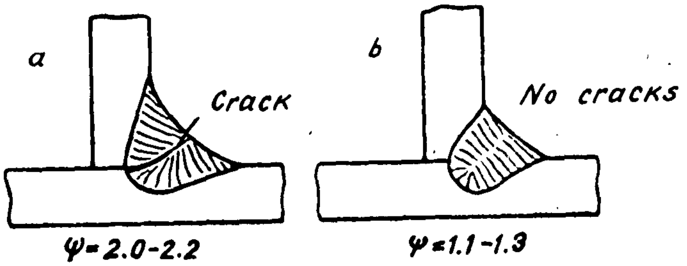

Fig. 119. Effect of shape factor ϕ on hot-crack stability of pure austenitic welds:
(a) electrode rod diameter 5mm; ϕ = 2.0—2.2: crack
(b) diameter 3mm; ϕ = 1.2—1.3: no cracks

Reducing the cross section and altering the shape of welds in order to eliminate cracking is very often resorted to in manual welding of austenitic steels and multipass welding under flux. It is true that in welding Kh23N18 (EI417) steel one sometimes succeeds in avoiding hot cracks in single-pass welding of metal 20mm thick. In small-size multipass welds of this steel cracks may appear in the successive layers required to fill in a Vee-groove in a butt joint. Austenitic steel welding practice often aims at obtaining a convex weld surface in preference to a concave one. This undoubtedly reduces the susceptibility of the weld to surface hot cracks. However, altering the shape of the weld does not by itself wholly eliminate the hazard of cracking. In fact, cracks may sometimes occur inside the metal (see Fig. 118) even though external ones do not appear. In welding under flux, lowering the weld's shape factor ϕ from 2—2.2 to 1.2—1.3 no doubt increases the hot-crack resistance of pure austenitic steels (Fig. 119). For this reason, in welding stable austenitic steels a thinner electrode rod of diameter 3mm, instead of 5mm, should be given preference.

As shown in works of TsNIITMASh (Central Scientific Research Institute of Technology and Machinery) stamping (peening) of edges results in a finer structure of the weld and enhances its crack resistance.

Similar results are obtained in multipass welding by peening each layer / 171, 189/. Regrettably, this method of combating cracks cannot be utilized when welding under flux. In effect, owing to the greater depth of fusion involved, the welded layer of metal is fused entirely and the intended effect of obtaining a finer structure is not achieved.

3. Fluxes and Electrode Rods for Automatic Welding of Chromium-nickel Austenitic Steels

Welding fluxes. In the early stage of automatic welding procedures for chromium-nickel austenitic steels, fluxes AN-5 /12/, FTsL-1 and FTsL-2 (Table 103) were applied.

Table 103

High silicon fluxes previously used in automatic welding of chromium-nickel austenitic steels

Brand of flux	Chemical composition in %								
	SiO_2	TiO_2	CaO	MgO	Al_2O_3	CaF_2	FeO	S	P
								Not over	
AN-5	50--52	-	26--30	8--14	Up to 2.5	5--7	Up to 1	0.10	0.10
FTsL-1*	27.5	10.0	28.5	17.2	12.8	-	4	-	-
FTsL-2*	36.0	-	32.0	16.0	6.0	8.0	2	-	-

* Calculated composition

It should be noted that a flux for austenitic steel welding has been devised in Poland, which closely approaches the composition of those indicated in Table 103; it contains: 35 % SiO_2, 18 % Al_2O_3, 32 % CaO, 10 % MgO, 6 % CaF_2 /86/.

Experience gained in the application of these fluxes revealed a number of shortcomings, chief among which being insufficient resistance of the weld to hot crack formation.

This disadvantage of acid fluxes makes itself felt more particularly in welding rigid constructions, and T-joints as well as in closed-butt annular welding of vessels; it is of less consequence in welding readily deformable work pieces and in fusing. In view of the drawbacks enumerated above, the high silicon fluxes have been replaced by the low silicon fluxes AN-26 and AN-23 (Table 104).

Table 104

Brand of flux	Chemical composition in %								
	SiO_2	CaO	MgO	MnO	Al_2O_3	CaF_2	FeO	S	P
							Not over		
AN-26	30-32	5.0--6.5	16--18	2.5--3.5	20--22	20--24	1.5	0.07	0.06
AN-23	17--20	10--13	12--15	8--10	20--23	22--26	1.5	0.07	0.06

These fluxes are incomparably better than those used in 1948-50. They guarantee a relatively low oxidation of chromium and ensure a comparatively high resistance of welds to the formation of hot cracks when used together with austenite-ferritic wires.

Judging by the data in literature, the switching from the use of high-silicon manganese-free fluxes to the use of those with a lower silicon dioxide content has been accomplished abroad too. At first the UM-20 flux, with a 50% SiO_2 content, was in use, while at the present the UM-80 flux, with a 38% SiO_2 and a 7% MnO content is recommended for welding of austenitic steels. The composition of these fluxes is given in Table 105.

Table 105

Fluxes used abroad for automatic welding of shromium-nickel austenitic steels /165, 225/

Flux brand	Chemical composition in %							Remarks
	SiO_2	TiO_2	CaO	MgO	MnO	Al_2O_3	CaF_2	
UM-20	50	--	30	10	None	4	6	Currently not used
UM-90	33	1	7	2	28	20	5	Used in USA, Fed. Rep. of Germany, and Austria
UM-80	38	--	22	10	7	15	6	

The UM-90 flux is employed together with the 29-9 type wire, with the addition of 4% molybdenum, in the form of ferromolybdenum, to the flux./226/. Molybdenum and chromium are added to the UN-88 flux; the composition of this flux is not known to us.

The low-silicon AN-26 and AN-23 fluxes, notwithstanding their advantages in comparison with the fluxes used formerly for welding of austenitic steels, are already inadequate to contemporary requirements. Being fairly suitable for welding of austenite-ferritic welds, these fluxes are, due to their oxidizing capability, contamination of the weld metal with silicate impurities, and a considerable passage of silicon into the weld metal, little suitable for welding of stable austenitic welds and of high-nickel heat-resistant alloys. The oxygen-free fluoride fluxes (Table 106) which form practically non-oxidizing slags during welding, are the most suitable for welding of austenitic steels and of hich-nickel alloys (cf. Ch. II).

Table 106

Oxygen-free fluoride fluxes used for automatic welding of austenitic steels

Flux brand	CaF_2	NaF	SiO_2	S	P
			Not nore than		
BKF-1	92	—	5	0.1	0.015
ANF-5 (former designation BKF-5)	75—85	15—25	2	0.08	0.015

A characteristic property of the BKF-1 flux is the shallow slag which it produces. As a result of this, the welds lack a sufficiently smooth transition into the base metal. The BKF-1 is a non-fused flux. It is obtained by grinding first-grade fluorspar (fluorite) from which the ferrous sulfide (pyrite) has been carefully removed. The ANF-5 is a fused flux, prepared from fluorspar (fluorite) concentrate or from chemically pure CaF_2 and sodium fluoride. Owing to the addition of NaF, the ANF-5 flux gives a less shallow slag. Welds made with the use of this flux have a good form, with smooth outlines of reinforcement. Because of the lower adhesiveness of the slag at solidification temperature of the weld pool the ANF-5 flux ensures a more complete elimination of hydrogen from the weld metal.

Electrode wires. The welding wire standard GOST 2246-51, which was in effect until May 1st, 1954, failed to take into consideration the demands which welding engineers placed upon electrode wires for welding of the austenitic 18-8 type steels. We have repeatedly emphasized that the presence of a ferritic component must be ensured in welding of this type of steels, which is necessary in order to avoid the formation of hot cracks as well as to guarantee the corrosion resistance. For realization of this requirement it is indispensable to have a welding wire of a corresponding composition, where the content of ferritizers is higher than that of austenizers. However, should one consult the GOST 2246-51, it will show that a limited content of ferritizers and, conversely, an excessively high content of nickel (up to 11%), is the characteristic feature of the 18-8 type austenitic wires.

The experience of utilizing the standard Sv-Kh (OKh18N9) type wires in the course of many years has shown that all too many melts had to be rejected due to the presence of cracks in the welds. It is known that the δ-phase hampers high-temperature mechanical processing of the 18-8, 25-12, and similar type steel ingots. The makers of austenitic wire therefore always strive to conduct the melting of these steels in such a way that the amount of the δ-phase is at a minimum. For this purpose the contents of chromium and silicon in the steel are reduced to the lowest limit permitted by the standard, while the concentrations of nickel and carbon are, on the other hand, brought up to the highest permissible limit.

A number of studies clearly indicate that the ratio of the Cr and Ni contents should not exceed 1.8, which means that at 18% of chromium the wire must have not less that 10% of nickel, according to the opinion of metallurgists. As a consequance of this the Sv-OKh18N9 type welding wire usually contained 10--11% of nickel, not more than 18% of chromium, and 0.2--0.3% of silicon. When this type of wire is used in welding, the weld has as a rule the least resistance to hot cracks. It is perfectly obvious that such a situation could not have been endured any longer; it was necessary to alter the ratio of the ferrite-forming components and of the austenite-forming admixtures in wires used for welding of the 18-8 type steels. A reduction of the sulfur content of the welding wire was also imperative. On the basis of experiments conducted within the recent years and of the experience of many plants, new components for wires employed in welding of the 18-8 type steels were proposed.

Some of the new wires are included in the GOST 2246-54, which became effective on May 1st, 1954. Their composition is given in Tbale 107, along with the composition of the 18-8 type wires manufactured as prescribed by the old standard and given here for comparison.

Experiments have shown that the Sv-OKh18N9S2 type wire is equally suitable for both automatic and manual welding of austenitic steels. It ensures a high resistance of the welds to hot cracks and non-susceptibility to intergranular corrosion when in the natural state.

The Sv-Kh18N12M type wire, made as specified by the 1951 standard, as well as the similar to it brand EI400 wire, also causes hot cracks in welds in a number of cases when the nickel content is near the upper limit (14%) and the silicon and chromium contents are near the lower limits (0.2--0.3 and 18%, respectively). The nickel content had therefore to be limited to 12%, and that of sulfur to 0.02%.

The use of welding wires given in Table 107 does not guarantee that welds resistant to intergranular corrosion under effect of critical temperatures will be obtained. The Sv-oKh18N9S2 type wire solves this problem only partially in those cases when the base metal pontent of the weld is at a minimum ($\gamma \leqslant 0.40$) and when its silicon content can be increased to 1.6--1.7%. When square-butt welding, $\gamma \geqslant 0.60$, as a result of which the silicon concentration in the 1Kh18N9T type steel welds, welded with the Sv-0Kh18N9S2 type wire, usuallyy does not exceed 1.2—1.3%.

The GOST 2246-54 provides for production of titanium and niobium alloyed wires for the purpose of obtaining stabilized welds (Table 108). According to the GOST these wires must be used only for manual welding (Table 108).

The titanium-containing Sv-1Kh18N9T type steel wire is emploed not only in manual, but also in electric gas welding. As a consequence of the development of oxygen-free fluxes it may also be used for automatic submerged-arc welding, in which case it is desirable for the titanium content of the wire to be not less than 0.7%.

The works of N. Yu. Pal'chuk, Yu. I. Kazennov, Ye. V. Sokolov, and S. V. Yunger have shown that the niobium-containing Sv-1Kh18N9B type steel wire may be used in manual welding only under the condition of additional alloying of the weld with silicon and molybdenum and of employment of coating of maximum basicity in order to escape hot cracks. For the same purpose L. M. Yarovinskiy and V. V. Bazhenov suggest introduction of metallic chromium or ferrochromium into the coating if the nickel content of the wire is more than 9%. Thus, for example, the Ts-11 type electrodes must be additionally alloyed with chromium, so that at

9--10% or 10--11% of nickel there would be 20% od 22%, respectively, of chromium in the weld. The Kh18N11B type steel wire cannot be recommended for flux welding of the 1Kh18N9T type steel, since even the use of oxygen-free fluxes fails to guarantee the absence of hpt cracks. Besides, its niobium content is insufficient for stabilization of square-butt welded welds.

Table 107

Austenite-ferritic electrode wires for welding of the 18-8 type steels, according to the All-Union State Standards (GOST) 2246-51 and 2246-54

Wire brand	Standard	Chemical composition in per cent							
		C	Mn	Si	Cr	Ni	Mo	S	P
		Not over						Not over	
Sv-0Kh18N9	Old	0.06	1.0--2.0	Not over 0.50	18.0--20.0	8.0--11.0	--	0.03	0.03
Sv-0Kh18N9	New	0.06	1.0--2.0	0.50--1.0	18.0--20.0	8.0--10.0	--	0.02	0.03
Sv-0Kh18N9S2	New	0.06	1.0--2.0	2.0 --2.75	18.0--20.0	8.0--10.0	--	0.02	0.03
Sv-Kh18N12M	Old	0.06	1.0--2.0	Not over 0.70	18.0--20.0	11.0--14.0	2.0--3.0	0.03	0.03
Sv-Kh18N11M	New	0.06	1.0--2.0	0.3 --0.7	18.0--20.0	10.0--12.0	2.0--3.0	0.02	0.03

Table 108

Electrode wires used to obtain intergranular corrosion resistant welds

Wire brand	GOST or TU (technical specification)	Chemical composiotion in per cent									
		C	Si	Mn	Cr	Ni	Ti	Nb	V	S	P
		Not over								Not over	
Sv-1Kh18N9T	GOST 2246-54	0.10*	0.30--0.70	1.0--2.0	18.0--20.0	8.0--10.0	0.5--0.8	--	--	0.02	0.03
Sv-1Kh18N9B	GOST 2246-54	0.09	0.3 --0.8	1.0--2.0	18.0--20.0	8.0--10.0	--	1.2--1.5	--	0.02	0.03
0Kh18N9F2S (EI606)	TU	0.07	1.3 --1.8	1.0--2.0	18.0--20.0	8.0--10.0	--	--	2.2--2.7	0.02	0.03
0Kh18N9FBS (EI649)	TU	0.07	1.3 --1.8	1.0--2.0	18.0--20.0	8.0--10.0	--	1.0--1.4	1.3--1.8	0.02	0.02

* In accordance with Paragraph 4 of the GOST 2246-54, the carbon content of the Sv-1Kh18N9T type steel wire may be limited to 0.05% C if valid reasons are given in the user's request.

It has been noted that Sv-Khl8N11B rod and the rod with an increased (up to 2.2) niobium content (suggested at the time by the Central Scientific Research Institute of Technology and Machinery) are not recommended for automatic flux welding. This rod does not make the welds resistant to hot cracks; and in welding on both sides when the Nb content increases in the weld and cracks might not appear, a decrease in the plastic properties of the weld is observed.

Austenitic-ferrite rods are the most suitable for welding of acidproof 18-8 steels; these rods provide for a complex additional alloying of the weld with chromium - silicon - vanadium OKhl8N9F2C (EI606), chromium - silicon - vanadium - niobium (OKhl8N9FBC (EI649) /Table 108/, and also Ti + Al(EI793)*.

Here as well as in foreign countries, rods with an increased content of ferrite-forming mixtures are lately being used in the welding of austenitic steels. Thus, for instance, a new American standard of 1954 planned a production of electrodes which provide for two-phase weld structure.

As has been mentioned earlier, in the USA, a 29-9 steel rod with or without niobium is used in flux welding, and in Germany -- 18-8 steel rod with 2 to 2/5% silicon /165, 225, 231/.

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* EI793 rod has been developed by the Kharkov Branch of the Scientific Research Institute of Chemical Machinery.

In conclusion it is necessary to note that in flux welding of austenitic steels, the output of high-quality welds is primarily determined by the properties of the rod. Therefore, in welding various austenitic steels the same flux may be used, while the selection of this or that type of rod will wholly depend on the type of steel and on conditions under which the rod will be used.(see Chapter 2, paragraph 1).

In Chapter 4, it was repeatedly emphasized that not austenitic-ferrite but pure-austenitic electrodes are necessary in welding stable austentic steels. It is necessary here that the silicon content in the weld be least and the carbon concentration be increased. As a result of welders' requirements, the silicon content of the standard welding Sv-Kh25N20 steel rod is decreased from year to year. However the rod mentioned above (Table 109) cannot as yet be considered useful. This rod has a great tendency to produce hot cracks in the welds, since the mere decrease in silicon content does not yet solve the problem.

TABLE 109

Variation in chemical composition of the standard rod for welding of 25-20 steels.

Year the standard was issued	Chemical composition in %						
	C	Si	Mn	Cr	Ni	S	P
	not more than					not more than	
1943	0.15	Not more than 1.0	2.0	24 - 27	19 - 21	0.030	0.030
1951	0.12	Not more than 0.7	1.0 - 2.0	24.0 - 27.0	19.0 - 21.0	0.03	0.03
1954	0.15	0.20 - 0.50	1.0 - 2.0	24.0 - 27.0	17.0 - 20.0	0.02	0.03

TABLE 110

Austenitic electrode rods for welding 25-20 steels.

Type of rod	Chemical composition in %								
	C	Si	Mn	Cr	Ni	S	P	V	W
		not more than				not more than			
Kh25N15G7	0.20 - 0.28	0.25	6.0 - 8.0	24.0 - 27.0	15.0 - 17.0	0.02	0.035	-	-
Kh25N15G7V3	0.18 - 0.25	0.25	6.0 - 8.0	24.0 - 27.0	15.0 - 17.0	0.02	0.035	-	3.0 - 4.0
Kh25N15G7F	0.18 - 0.25	0.25	6.0 - 8.0	24.0 - 27.0	15.0 - 17.0	0.02	0.035	1.5 - 1.9	-

On the basis of joint investigations by the Institute of Arc-Welding and the "Elektrostal' " plant, new types of rods are suggested for welding 25-20 steels. These rods have a low silicon content and an increased carbon content; their nickel is partly substituted by manganese; additional alloying with tungsten and vanadium is specified (Table 110). Due to its negative effect on the scale-resistance of welds, the rod containing vanadium must not be used in making constructions the welds of which operate at temperature more than 900°C.

The production of austenitic-ferrite rods having a large quantity of δ-ferrite and austenitic rods with low silicon content presents difficulties to metallurgists. In the first case, the hindrance lies in poor workability of the two-phase ingots, in the second -- the necessity of working melts in the induction furnaces on particularly silicon-free raw materials. Both difficulties can be eliminated, if the ingots for weld rods will be made according to a new method devised by the Institute of Arc-Welding /101/. This method provides for a melt without an arc with a consumable electrode in a copper water-cooling crystallizer. Following the favorable results of primary crystallization of the ingot and the uniform distribution of δ-ferrite, the ingots, cast by the new method, easily endure the hot mechanical process, rolling and forging. There are no difficulties at all in obtaining metal with a low silicon content by using a consumable electrode of weld silicon-free Sv-08 rod and by conducting the melting under silicon-free oxygen-free fluoride slag.

Earlier, it was mentioned that in a number of cases it is necessary to control the amount of δ-ferrite in the welds: in the case of welding heat-resistant steels, this is necessary to avoid intensive sigma phase formation of weld metal; in the case of welding acidproof steel, this is necessary in order to avoid decreasing the over-all chemical resistance in the nonacidifying media. Therefore, in many countries, great attention is given to the development of methods for evaluating the amount of σ-ferrite in weld rods, electrodes and fused metal.

The amount of ferrite can be determined by the ratio of the area occupied by the δ-phase in the field of vision at a magnification of 100 to 300 times (depending on the degree of structure refinement) to the entire surface of the microphotograph 9 x 12 or 13 x 18 mm. When the scale of "Zaporozhstal' " is used (Table 7), the amount of ferrite is thus determined, i.e., by the relationship between the areas of δ-phase and the entire microphotograph. At the present time, physical methods for evaluating the amount of δ-ferrite are being used: according to magnetic permeability of the sample when using a "ferritometer" or according to the intensity of the interference lines on the Debye crystallogram when using X-ray structural analysis. Incidentally, in Britain, a pocket ferritometer has been devised which by the needle deflection of the galvanometer makes it possible to evaluate the amount of ferrite directly in the welds without destroying them (Fig. 120).

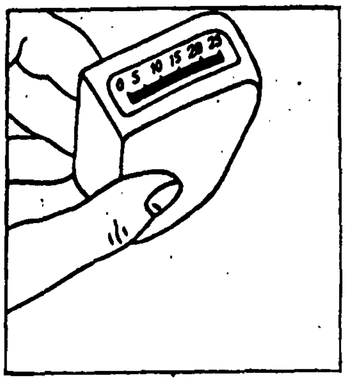

Fig. 120. Ferritometer for determining the amount of ferrite in welds of austenitic steels.

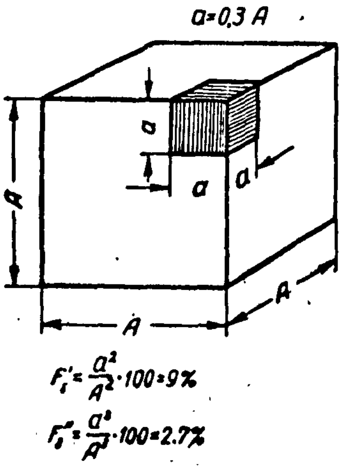

Fig. 121. Diagram explaining the difference in the indices of the amount of ferrite in the weld when using metallographic and physical magnetic methods.

In evaluating the amount of ferrite by means of the ferritometer, the gauge point according to the δ-phase is always considerably lower than the determination of the ferrite amount obtained by the metallographic method. Here, however, there is no contradiction. The metallographic method takes into account the relation of areas, and the magnetic method takes into account the relation of the volumes of the various phases. This can be seen from the following example.

Let the elementary volume of a metal 9cube of side A) contain ferrite of a volume a^3 (Fig. 121), where $a \approx 0.3\ A$. In the metallographic method, the amount of ferrite F'_δ is $(a^2/A^2) \cdot 100 = 9\%$, while in the magnetic method, $F''_\delta = (a^3/A^3) \cdot 100 \approx 2.7\%$, or three times less (Fig. 121).

4. Flux welding of the most widely used chromium-nickel austenitic 18-8 steels

In modern technology, the purpose of flux welding is to obtain austenitic-ferritic welds of 18-8 steels (1Kh18N9T, Kh18N11B, Kh18N12M2T, Kh18N12M3T, Kh18N9S3 etc.) as well as single-phase welds of 25-20 steels (Kh23N18, Kh23N23M3D3 and other).

Welding of chromium-nickel-titanium 1Kh18N9T steel. AN-26 or ANF-5 fluxes are used in automatic and semiautomatic welding. The brands of electrodes, the mode and the techniques of welding (the order in which the welds are made) are chosen depending on the specific conditions in which a welded structure is required to perform.

However, regardless of the conditions for which a given structure is designed, the electrode should have the minimum admissible amount of nickel, preferably not more than 8 to 9 per cent. The use of high-nickel electrodes (for instance 15 to 20 per cent) cannot be tolerated by any means. Fig. 122 shows as an example the macrograph of a double butt weld of 30 mm thick 1Kh18N9T steel. One of the welds (b) is erroneously welded with a 25-20 steel electrode. This weld is of a purely austenitic structure and is affected by hot cracks. The other weld (a) is made with a Sv-0Kh18N9 electrode; it is free from cracks and has an austenitic-ferritic structure.

Fig. 122. Macrograph of a 30 mm-thick automatic butt weld of 1Kh18N9T steel:

a) crack-free weld, made with a Sv-0Kh18N9 electrode;

b) weld with crack, made with a Sv-Kh25H20 electrode.

In the welding of structures of the first group (see Chapter II, Section 1), the welds in their natural state are required to be resistant to intergranular and overall corrosion. Occasionally, this requirement can be met by using the standard Sv-OKh18N9 or Sv-Kh18N11M electrode brands, provided that the welds are free of hot cracks. Still better results are obtained by using the high-silicon electrode, brand Sv-OKh18N9C2. When electrodes of the said brands are used in welding, the welds that are supposed to face an aggressive medium should not be again subjected to welding heat. This means that in the manufacture of most vessels, where the inner welds are the working ones, it is essential that the inside be welded after the outside welding is completed. Here, the first weld has to be welded in free suspension which is extremely inconvenient because avoiding gaps between the butts is not possible. Hence, high-silicon Sv-OKh18N9 electrodes and others of this type should be primarily applied in single full-penetration welding or, on condition that the working welds can be made first, in double welding also, as well as if the welded structures are required to perform in moderately aggressive media. In V-butt welding, OKh18N9S2 electrodes can be used with good results.

In the welding of structures of the first group, the best results are obtained with OKh18N9F2S (EI606) electrodes, the use of which ensures additional alloying of the welds with vanadium and silicon, or using Sv-1Kh18N9T or OKh18N9TYu electrodes in combination with flux ANF-5.

When welding structures of the <u>second group</u> the welds of which perform in aggressive liquid media, and are periodically exposed to critical temperatures, the vanadium-silicon-bearing electrode, brand

OKh18N9F2S (EI606) may be successfully applied. The vanadium-silicon-niobium-bearing electrode, brand OKh18N9FBS (EI649) may be used also, provided thorough checking for hot cracks. Electrode, brand Sv-1Kh18N9T or OKh18N9TYu may be used in combination with flux ANF-5.

It should be emphasized that if OKh18N9F2S, OKh18N9FBS, Sv-1Kh18N9T electrodes are used in welding, the order in which the working and the backing welds are made is hardly of importance and does not practically effect the corrosion resistance of the welds. If the working medium is highly aggressive, it is desirable that the weld joints be stress-relieved by stabilization annealing, no matter which of the three said electrodes is used. This eliminates the posibility of knife-line corrosion and of corrosion under stress.

For the welding of structures of the third group which perform continuously at critical temperatures, the brand of electrode should be chosen on the basis of the following security requirements: 1) corrosion resistance in liquid and in gaseous media, 2) resistance to scaling, 3) stability of the microstructure and prevention of δ-phase $\rightarrow$ σ-phase regeneration. Alloying systems that would in equal measure meet each of the said requirements are not and cannot be.

If it is the second and the third requirements that are to be primarily considered, the high-chromium-silicon Sv-0kh18N9 or Sv-0Kh18N9S2 electrodes should be used. In this kind of welding, the immunization period of the weld is not too long, while the increased content in silicon and chromium has a beneficial effect on the scaling resistance of the weld. The relatively small amounts of δ-phase that are to be found in welds made with Sv-OKh18N9 electrodes, ensures adequate resistance to σ-phase formation under exposure to critical temperatures. Owing to chromium and silicon, welds made with Sv-OKh18N9S2 electrodes are more resistant to scale formation but their higher δ-phase content makes them somewhat more susceptible to σ-phase formation.

If corrosion resistance is required in the first place, the OKh18N9F2S (EI606), OKh18N9FBS (EI649) and 1Kh18N9T electrodes should be used. Since tests revealed that vanadium slightly reduces the scaling resistance of welds, the last two electrodes should be given preference. Here, it should be kept in mind that niobium, as incidentally vanadium, chromium and silicon, contribute to σ-phase formation in the welds. It is irrelevant in which order the structures of the third group are welded.

If 1Kh18N9T steel is used as heat-resistor, the electrode brands Sv-Kh18N11M (after a thorough hot-crack check) and Sv-1Kh18N9T may be applied.

The industrial methods by which 1Kh18N9T steel of various thicknesses is flux-welded are to be found in works /125, 40, 111 and 142/ as well as in the pamphlet of this author.

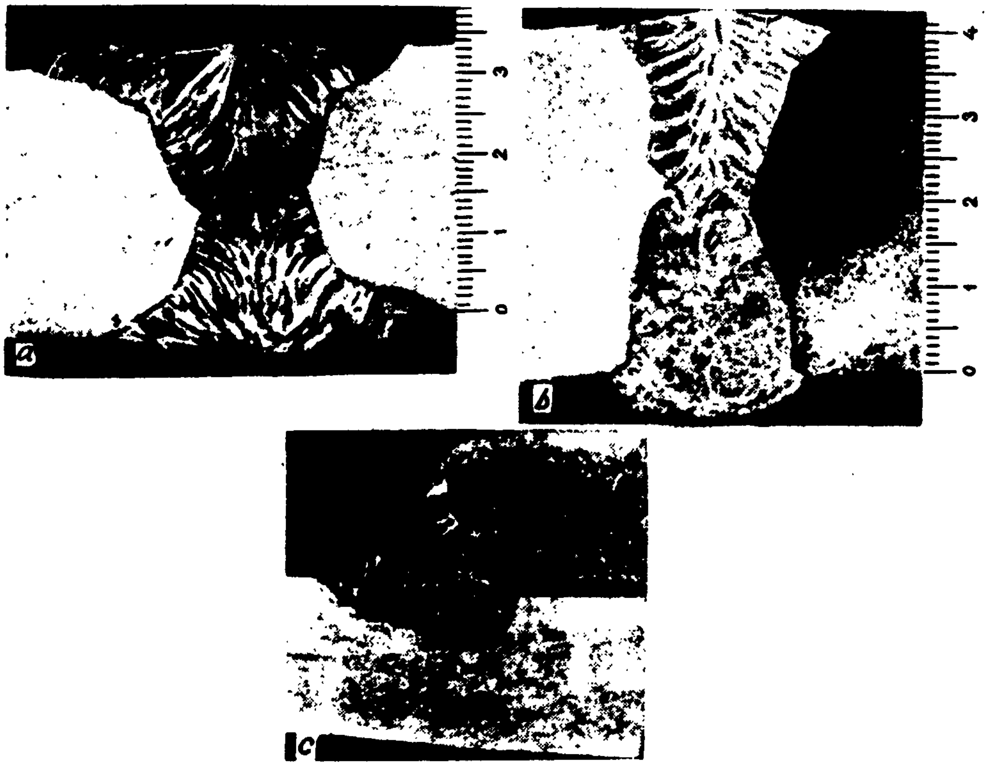

Fig. 123. Macrographs of automatic austenitic-steel welds: a) double square butt weld of 35 mm thick 1Kh18N9T steel, b) same for steel, 40 mm thick, c) fillet weld of 12 mm thick Kh18N12 steel.

welding of thick 1Kh18N9T steel. However, the situation changes in the case of multi-pass V butt welding. When using wire with a high ferrite content, alloyed with titanium, vanadium and niobium, embrittlement of the weld is possible already during the welding process. A decline in the plastic properties of the weld is all the more possible as a result of the effect of the service temperatures. Therefore, multi-pass welds must be made according to the following procedure. Almost all sections of the V butt must be filled with an Sv-0Kh18N9 or Sv-0Kh18N9S2 wire, while 1Kh18N9T, EI606 or EI649 wires are used only for depositing the upper layer facing the aggressive medium. This eliminates the danger of embrittlement of the weld metal and preserves its resistance against the aggressive medium. This is equally true for other methods of arc welding of austenitic steels of great thickness.

Welding of Kh18N11B chromium-nickel-niobium steel. This steel differs from 1Kh18N9T steel not only by the fact that titanium is replaced by niobium, but also by a higher nickel content. Welding of chromium-nickel-niobium steels, which have a tendency to cracking not only in the weld metal but also in the zone adjacent to the weld, involves certain difficulties; more complex procedures are in many cases required.

Wire having the same composition as the base metal must never be used in welding of this kind of steel. When selecting an electrode wire for welding these steels it must not be forgotten that niobium intensively promotes hot cracking, and this detrimental effect of niobium becomes extremely stro as the nickel and carbon content increases. Kh18N11B steel must be welded under ANF-5 flux; the use of AN-26, AN-23 and similar fluxes is not recommended.

Sv-0Kh18N9 or Sv-0Kh18N9C2-type wire can be used in welding of structures of the first group, like in welding of 1Kh18N9T steel. However, V butt welding becomes mandatory in this case. Single-arc welding is not recommended for crucial structures, even in the case of V butt joints. Twin-arc welding with thin (2 to 3 mm diameter) separated electrodes is to be applied. Welding can also be done with square-butt joints, but the mentioned wires must not be used in this case. In square-butt welding of Kh18N11B steel weldments it is necessary to use a 0Kh18N9F2S (EI606) chromium-nickel-silicon-vanadium wire. 0Kh18N9FBS (EI649) wire containing niobium is not necessary here, since the weld becomes alloyed with niobium through the passage of the latter from the base metal melted by the arc into the weld pool.

We already mentioned that in square-butt welding of Kh18N11B steel containing up to 1% niobium, up to 0.5 to 0.6% niobium passes into the weld. It is desirable to apply preliminary heating in welding of Kh18N11B steel. Heating to not less than 600 to 650°C is required in welding with an Sv-OKh18N9 type wire. However, if welding is done with a wire containing vanadium, the heating temperature may be reduced. In many cases, heating of Kh18N11B steel is not necessary at all when using an EI606 wire. An EI793 wire may be used /236/.

According to experimental data, a 25-8-type wire can be used in square-butt welding of Kh18N11B steel, and according to data from the literature a 29-9-type steel wire also. It must be pointed out that chromium-nickel steels containing niobium are more often used for constant service at high temperatures than 18-8-type steel containing titanium. The necessity of ensuring resistance of the weld against formation of the sigma phase therefore becomes first and foremost. Proceeding from this idea, it cannot be considered expedient to use welding wires with an elevated chromium content (25-8 and 29-9). The same is true -- though to a lesser degree -- for OKh18N9F2S (EI606) wire containing vanadium. In a number of cases it is therefore desirable to use an Sv-OKh18N9S2-type wire in welding of structures of the third group, in spite of the necessity of applying more complex procedures (V butt welding, twin-arc welding, heating, welding with "split" electrodes).

Hot cracking in the vicinity of the weld is a specific peculiarity in welding of chromium-nickel-niobium steels. The danger of this kind of hot cracking can be somewhat reduced by heating the part to be welded and by welding at regimes characterized by a low energy per unit length. Calking of the edges may also prove useful. The appearance of high-temperature ferrite in the base metal at the fusion line prevents crystallization cracking in the vicinity of the weld.

It is not without interest to point out that flux welding of chromium-nickel-niobium steels had been advertised in the foreign literature a few years ago, and the question of hot cracking in welding of these steels was not raised until the end of 1951. In August 1953 an American journal published a comprehensive paper on welding of austenitic 18-8-type steels containing niobium (AISI-347). The paper stated the results of inquiries made at the 23 largest enterprises producing weldments of these steels. It turned out that all plants without exception encountered hot cracking in welds and 40% of the plants encountered cracking in the vicinity of the weld in 18-8 steel containing niobium. The paper indicates that automatic flux welding of this kind of steel is practically not done, owing to technical difficulties due to the tendency to hot cracking. Hot cracking is very often observed in argon arc welding of AISI-347 steel also. In this connection, it must be emphasised that flux welding of austenitic chromium-nickel-niobium steel has been introduced in the industry of the USSR.

Welding of Kh18N12M2T and Kh18N12M3T chromium-nickel-molybdenum-titanium steels. When welding Kh18N12M2T steel, there is, as a rule, no hot cracking in the weld, in spite of the relatively high nickel content. This is explained by the beneficial effect of molybdenum. There is more molybdenum in Kh18N12M3T steel, but hot cracking in welding occurs much more often, since these steels are usually produced with the highest permissible nickel content, i. e., up to 14%, whereas the nickel content in Kh18N12M2T steel seldom exceeds 12%.

It is advisable to inspect both types of steel for hot cracks just before it is used in production. If the test is negative, the steel can be sent for welding by the conventional technological methods without preparation of the edges. But if it is found during the test that the steel is likely to crack when standard wire and fluxes are used, there has to be a complex system of welding with preparation of the edges, preheating, and so on. According to literature on the subject, the welding of chromium-nickel-molybdenum austenitic steel requires preliminary and concurrent heating up to 315°C /202/.

Kh18N12M2T and Kh18N12M3T steels are best welded under ANF-5 flux. The choice of the electrode wire is determined by the nature of the agressive medium. If the medium is not particularly agressive, Sv-OKh18N962 wire can be used. In welding without edge preparation there will not be more than 1.2 to 1.4% molybdenum in the weld, provided the steel contains a minimum of 2% of this element. It is therefore better, of course, to use Sv-Kh18N11M wire, but only after a thorough check for hot cracks.

If Sv-Kh18N11M welding wire is used, and there is repeated action by heat, the weld will not be sufficiently resistant to intercrystallite corrosion. Metal parts belonging to this category, which operate in sulfuric acid, should be welded with Sv-Kh25N13 or Kh18N11MT wire. The first of these is manufactured as GOST (National

Soviet Standard) 2246-54, but the wire can only be used when it contains not more than 0.08% carbon (up to 0.12% under GOST), not more than 12.5% nickel (from 12 to 14% under GOST), not less than 25% chromium (23.0 to 26.0% under GOST) and not less than 0.6% silicon (0.3 to 0.7% under GOST).. The second wire is experimental. In composition it corresponds to the standard Sv-Kh18N11M wire, differing only in that it contains titanium (0.5 to 0.7%). When welding with Sv-Kh25N13 wire, stabilization of the weld is achieved through the increased chromium content (20 to 21%), given a molybdenum content of 1.2 to 1.4%. In welding with Kh18N11MT wire the stabilizing elements are titanium and molybdenum. Good results are obtained by the use of the experimental wires made from Kh25N12T and Kh20N12M3T steel.

If the steel is being used in an oxidizing medium, 0Kh18N9FBS (EI649) wire must be used.

In cases where the steel is being used as a heat-resistant material, Sv-Kh1811M wire should be used. If the steel is operating as a scale-resistant material, the welding should be carried out with Sv-Kh18N9S2.

Welds made of Kh18N12M2T steel have an extreme tendency toward knife-line corrosion (Fig. 101 e). Hence, in all cases when possible, welded joints intended for use in agressive media should undergo stabilizing annealing.

When Kh18N12M2T and Kh18N12M3T steels are welded, under no circumstances should welding wire made from chromium-nickel-molybdenum-niobium steel type Kh16N12M3B (EI403) be used on account of the danger of hot cracks.

Welding of Kh23N18 chromium-nickel steel (Fig. 124). Great difficulties are encountered in welding Kh23N18 steel because welds must be obtained that possess a purely austenitic single-phase structure. Standard Sv-Kh25N20 wire is inadequate for welding Kh23N18 steel. Satisfactory results can be obtained using ANF-5flux, combined with the wires specified in Table III, characteristic of which is a low silicon and higher carbontcontent and manganese, substituted in part for nickel. The wire diameter must not exceed 3-4 mm. When welding this steel, it is, however, of particular importance that the welding wire be thoroughly checked for hot cracks. At the present time, a careful sorting out of austenitic electrodes as to the hot-crack resistance of the weld metal is the only reliable method to obtain good-quality austenitic-stable steel welds. In many countries, welders have come to this conclusion. There exsist a mumber of methods to check austenitic electrodes for hot cracks. The simplest and most effective one is the welding-on of multi-layer beads and welding of T-specimens. In the manufacture of heavy-duty structures, it is general practice to test austenitic electrodes on specimens duplicating the actual welded joint [183].

In welding thick Kh23N18 steel, preheating to 150-200° may prove to be beneficial. In multi-pass welding of a thick metal, overheating should be avoided, to prevent hot cracks in the vicinity of the weld. A peculiar feature of the single-phase welding of austenite-stable steels is the danger of the occurrence of hot cracks in craters and tack welds. 25-20 steel structures, hence, should be assembled using austenite-ferrite electrodes made, for example, from Sv-Kh25N13 wire. In many cases this wire may be used in Kh23N18 steel flux welding.

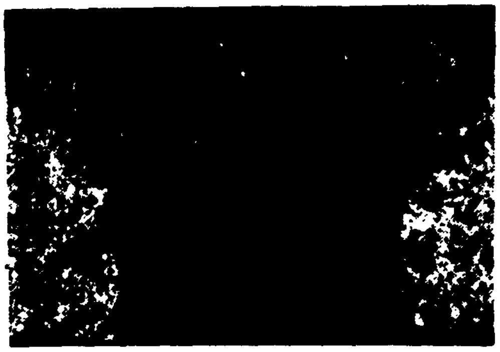

Fig. 124. Macrosection of 6 mm-thick Kh23N18-steel single weld.

Welding of chromium-nickel-molybdenum-copper Kh23N23M3D3 steel. This steel is used as an acidproof material in aggressive media which may cause intergranular corrosion. Welds of this steel must possess a purely austenitic structure. Out of the variety of the methods of increasing the hot-crack resistance of such welds only the following may be used: 1) reduction of silicon content; 2) the use of oxygenless fluxes. Increase in carbon content in the weld, though proved to be quite adequate in welding scale and heat-resistant steels, may not be used in this case in view of the danger of a steep drop of the corrosion resistance of the weld. For the same reason and because of its negative effect on the hot-crack resistance of welds, substitution of manganese for nickel must be carried out with great caution. In welding Kh23N23M3D3 (EI533) steel, it is recommended to use a wire of the following chemical composition: up to 0.06% C, up to 0.2% Si, up to 1% Mn, 22-25% Cr, 20-22% Ni, 2.5-3.5% Mo, 2.5-3.5% Cu, no more than 0.02% S and 0.03% P. ANF-5 flux must be used. We cannot accept AN-26 flux, recommended for use in welding steels similar to Kh23N23M3D3 (EI533) steel [13]. In welding Kh23N23M3D3 steel, single single-pass butt welds should be avoided whose back-side bead is wider than the weld itself. Transverse hot cracks on the side of the back-bead may occur in these welds. 3-4 mm D-wires must be used.

In copper-backed welding, it is necessary that there be a thin flux layer between the edges and the backing, to pretect copper from welding, and to prevent the occurrence of copper in the weld and the formation of hot cracks. This is revelant to all austenitic steels.

5. FLUX WELDING IN THE PRODUCTION OF EQUIPMENT FROM CHROME-NICKEL AUSTENITIC STEELS

The advantages of flux welding - its highest productivity, the highest quality of welds, the automation and semiautomation replacing the hard manual labor, the economy of power and welding materials, etc., which make it superior to other modern methods of arc welding, are generally known. These advantages are particularly prominent in austenic-steel welding.

Economically, the substitution of automatic and semiautomatic flux welding for austenitic-steel manual welding is extremely effective, primeraly, due to the simpler and cheaper preparation of the objects for welding. It is known that mechanical processing of austenitic steels is greatly impeded by their strikingly high viscosity and susceptibility to hardening. In stainless-steel manual welding, the preparatory treatment of edges is indispensable, beginning with a 3-5 mm thick metal. Basically, the preparation of edges must necessarily be performed using mechanical techniques. This makes the production of austenitic-steel items much more costly. In automatic flux welding, one may dispense with the preparation of edges even when a metal of practically unlimited thickness is welded together: steel, up to 40-50 mm thick, can be welded with the arc in lower position, and thicker metal using electro-slag welding. [114] . Basic, most frequently met, austenitic steels, for example, the chrome-nickel-titanium brand, can be welded without a preparatory treatment of edges.

The experience of oil and chemical machine-building plants which have extensively introduced the practice of submerged-arc welding in the production of domestic austenitic-steel machinery coming under the State Mining Supervisory Organization has shown convincingly that the reduction in the machining of these steels, brought about by the switch to submerged-arc welding, has resulted in tremendous savings.

Another source of economy is reduction in the consumption of costly austenitic welding wire.

The technological advantages of machine-arc welding of austenitic steels are especially great. Besides the generally known advantages which have been mentioned, there are other features of this method of welding which are not of importance in welding ordinary steels. By this is meant the standard of the shaping of welds, the state of the surface and also the state of the surface of the base metal near the weld. Machine-arc welds, as distinct from manual-arc welds, have a smooth, finely-flaked surface with gradual transition to the base metal. This fact greatly improves the over-all chemical resistance of welded joints and makes it easier to remove the products of corrosion in machinery which is in operation. The absence of spatter on the surface of the base metal reduces the danger of centers of corrosion arising to a minimum. According to data obtained by Yu. I. Kazennov, A. N. Krutikov and others (see coll. No. 20 of Scientific Research Institute of Chemical Machinery), and also data from N. M. Nikitinykh*, hot cracks are often detected in the spatter when

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* Private communication

austenitic electrodes are used in welding. We are not concerned here with such well-known advantages of submerged-arc welding as consistency in the quality of the welds, the absence of defects found specifically in manual-arc welding, such as non-fusion in the weld root and along the edges, macroinclusions of slag, and so on.

Of great importance also is the fact that in automatic and semiautomatic welding under flux a weld can for practical purposes be made with one unit from beginning to end without interruption. In manual-arc welding the weld consists of separate sections, the beginning and end of which are often spoiled by defects, compared with the central section. Each time the manual-arc welder changes an electrode, he subjects the end of the completed weld to the repeated action of heat; this cannot but have an adverse effect on the corrosion resistance of the weld. In submerged-arc welding the number of these sections can be reduced to a minimum, and this helps to increase the over-all chemical resistance of the welded machinery.

6. Electro-slag Welding of Austenitic Steels /248/.

A new method for the single-pass welding of thick metal -- electro-slag welding -- is based on use of the heat generated by an electric current as it passes through the slag which acts as a conductor. Electro-slag welding is described in detail in a monograph published by the Institute of Electric Welding /114/.

When applied to austenitic steels the process has particularly clear advantages, such as high efficiency, a minimal tendency on the part of the weld metal and the region around the weld to form cracks, the absence of zonal liquation, the possibility of welding metal of practically unlimited thickness without separating the edges, only slight warping as a result of the symmetrical distribution of heat with respect to the welded edges, and so on. At the same time, electro-slag welding, when applied to austenitic steel, has the disadvantage of causing appreciable overheating of the metal in the region of the weld. Austenite steels are less inclined to produce a larger grain than conventional structural steels. As a result, however, of the high degree of overheating and delay in cooling, electro-slag welding may produce a tendency toward knife-line corrosion in austenitic steel. A definite method of preventing this danger is the heat treatment of welded parts immediately afterwards, but this is not always possible in actual practice. Hence technological methods must be sought in order to prevent overheating and delayed cooling. Welders resort to welding with transverse shifting of the electrode in the gap between the edges being welded. The use of the oxygen-free fluxes BKF-1 or ANF-7, which are extremely good conductors in the molten state, is very effective. On account of this remarkable property of fluxes composed of calcium and sodium fluorides, a steady electro-slag process can be maintained at a welding-arc voltage of only 20 to 22 v, which is half as much as the voltage required when welding with AN-26 flux. The drop in the voltage results in a considerable decrease in the overheating and, consequently, less danger of knife-line corrosion of the weld and intercrystallite corrosion in the region.

In choosing electrode wire or sheets for electro-slag welding of austenitic steels, one should be guided by the same considerations as in carbon-arc welding.

7. Certain Aspects of the Gas-Shielded Arc Welding of Austenitic Steels /250, 256/.

As is known, gas-shielded arc welding in an atmosphere of argon and helium is used extensively for the production of parts made from thin sheets of austenitic steel, and also in welding butts with and without bends in tubing. When welding with a non-consumable electrode (tungsten) it is not possible to control the composition of the weld metal to a great degree. If the welding is done without filler wire, the weld metal is for practical purposes no different from the metal being welded in chemical composition. If the steel being welded has a one-phase structure with a large austenite concentration, the formation of hot cracks is inevitable (Fig. 72 a). The appearance of these cracks is greatly facilitated by niobium. As a result, 18-8 steel with Ni is in many instances considered undesirable, and is replaced by steel containing titanium.

This is in fact what is done in American aircraft production. In argon-arc welding with a consumable eledtrode it is possible to regulate the composition of the weld metal to a great extent. All the laws of behavior described in this book with regard to the positive rôle of the second phase in the case of two-phase welds, and of oxygen in the case of one-phase welds, apply fully to gas-shielded arc welding as well.

In view of the high price of argon and helium, welding in an atmosphere of carbon dioxide is being developed at a fast rate. According to literature on the subject, this type of welding has been successfully applied in the production of machine parts from heat-resistant nickel-base alloys, which are known as Nimonic and Inconel.

As shown by the experiments of Rothschild in the USA and our own experiments, carried out in collaboration with I. N. Rublevskiy, carbonizing of the welds is observed during the welding of austenitic steels. Rothschild has drawn the conclusion on the strength of this that carbon dioxide is unsuitable for use in welding austenitic acid-resistant steels. Since the transition of the carbon dioxide does not exceed 0.03 to 0.04% C, it may be considered that this conclusion is somewhat hasty. If it is possible to make an automatic weld containing 0.10 to 0.12% in 18-8 steel by the submerged-arc process, and that weld is fairly corrosion-resistant, there is no reason to believe that the same thing cannot be done in gas-shielded arc welding as well.

One has only to make electrode wire of the requisite composition. As regards stable austenitic heat-resistant and scale-resistant steels, welding in carbon dioxide may seem more hopeful since a certain carbonizing and oxidation of the hydrogen and silicon help to prevent hot cracks. Our experiments have shown, for example, that it is easy to weld pure austenitic welds in such one-phase steels as Kh23N18, EI533 etc. in CO_2 without hot cracks. On account of the carbonizing a high degree of heat-resistance is achieved during the welding of steel in CO_2 (see p. 288).

It may be considered that in the manufacture of parts from thin sheets of austenitic steels, and also in assembly welding when the part can be approached from different directions, gas-shielded arc welding, including carbon dioxide welding, will be further developed and will replace manual bare-arc welding. In the production of thick austenitic steel parts it is better to use submerged-arc welding. At the present time, in view of the production of oxygen-free fluxes, automatic shielded-arc welding can easily compete with gas-shielded arc welding in the inert gases argon and helium.

It should be stressed that when applied to acid-proof steels, CO_2 welding has the following disadvantages: spattering (up to between 10 and 12%) leading to the appearance of potential centers of corrosion, and the formation of an oxide film on the weld surface difficult to remove. In combined slag-gas shielded welding the oxide film is not formed. For this, 30 to 40 g of fluoride flux ANF-5 per 1 lin. meter of the weld have to be added to the arc zone during welding in carbon dioxide.

CONCLUSIONS

1. Chromium-nickel austenitic steels (acid-, scale- and heat-resistant) are employed under varied temperature conditions, ranging from very low temperatures up to 1,100° and over, and are subject to the action of diverse aggressive media. Depending on the composition and properties of the steel and on the specific operating conditions of the constructions concerned, welded seams of austenitic steels must satisfy certain requirements beyond the customary demands of strength, plasticity, and absence of cracks and porosity. In consequence, there is and can be not one (welding) technology common for all austenitic steels. Moreover, even in constructions from austenitic steel of the same brand welding technology must take into account their purpose and conditions of exploitation.

2. Theorization of experimental data has proved that there is a direct dependence between the nature of the primary crystallization of the weld pool, the primary microstructure of the weld metal on an austenitic steel and its essential properties, viz., stability against crack formation and resistance to structural (intergranular) and over-all corrosion.

3. It is shown that the weld metal on austenitic steel differs distinctly by its properties and the nature of its microstructure not only from rolled steel, but also from ingots of identical composition. Therefore, the opinion widely held in the literature that in welding austenitic steels one must aim at obtaining

a weld metal possessing the same chemical composition and structure as the welded steel is erroneous.

It was found that, in many cases, in order to obtain welded seams satisfying set requirements they have to be complexly alloyed with elements that are not present in the base metal, preventing at the same time enrichment of the seam with admixtures contained in the welded steel.

4. Stress is laid in the metallurgical literature on the necessity for preventing ~~to prevent~~ a two-phase structure in austenitic steels; it is held that the ㉕ ㉘ α(δ)-phase (ferrite) ~~is~~ definitely impairs the fitness of austenitic steels for engineering purposes, impeding its mechanical heat~~-~~treatment, such as forging, rolling and piercing, and lowers their mechanical properties at both low and high temperatures; ferrites, it is believed, cause a flow-line structure in rolled steel and lead to anisotropy. Until ~~Up to~~ recently, the above-cited views about an alleged harmful effect of the ㉕ ㉘ α (δ)-phase were also supported in the literature on welding.

lpha
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The present work proves the fallacy ~~fallaciousness~~ of the mechanical application of ~~mechanically applying~~ to the domain of welding technique the laws that pertain to metallurgy. It is shown that welded seams with a two-phase structure are decidedly superior to single-phase austenitic seams in respect to several essential characteristics (such as stability against cracks and corrosion) and that in the overwhelming majority of cases the primary ferrite δ-phase) is a most desirable structural component of the weld metal on austenitic steel.

5. It was established that the influence of welding conditions and of the system of alloying ~~upon~~ on the properties of the weld metal (mechanical properties and resistance to cracks and corrosion) must be viewed from the standpoint of their action on the nature of the primary microstructure of the welded seam. All alloying admixtures can be divided, according to their influence on the primary structure of the seam, into two groups, viz.: austenitizers and ferritizers.

6. Experiments conducted on the basis of A.A. Bochvar's theory of the hot brittleness of cast metal made it possible to establish a direct interdependence between the type of phase diagram, according to which crystallization of the weld pool proceeds, and the stability of the weld metal against hot crack formation. Certain general laws were formulated, which enable us to evaluate the effect of any element whatsoever on the formation of hot cracks in austenitic steel welding.

7. It is known that in the practice of austenitic steel welding use is made of electrode coatings and fluxes which produce slags of a more or less uniform type. A study of the effect of certain gases (hydrogen, oxygen and nitrogen) on the primary crystallization of austenitic steels helped to clarify to a

considerable extent the mechanism of the positive action of the basic slags on the properties of the seams. It was demonstrated that the modifying effect of the basic slags is linked with the ratio of the amounts of oxygen and nitrogen introduced into the weld pool.

9 — 8. An essentially new system of oxygen-free fluxes has been developed for the automatic welding of austenitic steels. These fluxes, which are based on fluorides of alkali- and alkali-earth metals, form virtually neutral slags after melting. Oxygen-free fluxes, it was found, do not enter into metallurgical interaction with liquid metal, do not oxidize the weld pool and do not prevent the latter from assimilating all the alloying elements, including also those having a ~~close~~ strong affinity ~~to~~ for oxygen, which are contained in the base metal and in the welding rod.

The employment of fluoride fluxes introduces fundamental changes into the metallurgy of welding high-alloy steels, making it possible with the aid of the electrode metal to alloy the seam with titanium, aluminum, zirconium, boron, etc.

9. Direct experiments brought out the fact that hydrogen is capable of ~~provoking~~ inducing hot cracks in purely austenitic welded seams; oxygen, by contrast, can prevent the formation of these cracks by its ability to oxidize hydrogen, silicon and sulfur.

It was established that in order to enhance the stability of purely austenitic seams against hot cracks it is mandatory to reduce the silicon content, increase the carbon concentration, replace part of the nickel by manganese and endeavor to achieve a finer primary structure of the seam, unless this should lead to increased sigma-phase formation in it.

10. Experimental studies disclosed the effect of the specific structure of a welded seam (microscopic liquation, finely dispersed disorientated δ ㉘ -phase distribution) and of the complex stressed state upon the rate and nature of development of diffusion processes in the weld metal. For the first time, it was proved that, owing to the above-mentioned characteristics of the weld metal on austenitic steel, the diffusion proceeds tens of times faster than in rolled steel. This enabled us to recommend and make practical use of an immunizing heat-treatment (stabilizing tempering) of welded constructions as a reliable means for improving the quality and prolonging the life of such constructions. On the basis of experimental and theoretical data, we recommend a new means of warranting the stability of welded seams against intergranular corrosion by creating a complex alloyed

delta

ferrite in the seam. It is shown that in the presence of separated regions of complexly alloyed chromium-rich ferrite in the austenitic base of the seam metal this stability is attained even if carbide-forming elements (titanium and niobium) are completely absent. The theory is advanced that the positive role of the δ-phase is linked to the localization of chromium impoverishment in the austenite due to the opposing diffusion processes developing in carbon (from the austenite into the ferrite regions) and chromium (from within to the periphery of these regions).

12. We investigated the conditions of origination and development of corrosion occurring along the line of fusion (so-called knife-line corrosion) in welded joints of acid-resistant steels of the 18-8 brand, containing titanium and niobium, in ~~both~~ oxidizing and in nonoxidizing media. It was found that stable austenitic steels of the Kh23N23M3D3 brand are subject to transcrystallite corrosion in sulfuric acid due to compression deformation provoking tensile stresses near the surface of the steel or weld.

13. It is shown, for the first time, that niobium provokes hot cracks during flux welding of austenitic steels of the 18-8 brand. It was established that the adverse effect of niobium can be fully neutralized by appropriate additional alloying of the weld metal with ferrite-forming elements --in the first place silicon and vanadium -- as well as by limiting the concentrations of nickel, carbon, oxygen,

nitrogen and sulfur.

14. Theorization of industrial experience and experimental work enabled us to devise and recommend for manufacture a new type of welding rod with a higher content of ferrite-forming admixtures, to be used for constructions from steel of the 18-8 brand.

. . . These rods, in combination with the basic slags (fluxes and electrode coverings) permit of solving the problem of arc-welding of the prevalent 18-8 brand austenitic steels.

15. There were developed low-silicon austenitic electrode rods which, in combination with oxygenfree fluoride fluxes, ensure adequate stability of single-phase austenitic welds againdt the formation of crystallization cracks in the welding of steels of medium thickness.

16. It was established that the slight carbonization of the weld metal(by 0.02 to 0.04 %) occurring during austenitic steel welding in CO_2 substantially increases its heat-resistance in the presence of Ti, Nb, W, and M_o in the weld.

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In the present book, we have dealt in a general way with the metallurgical and technological problems of welding austenitic steels of the basic types and the problems pertaining to the stability of welded joints of these steels against corrosion. Ways and means have been indicated which make it possible to exert an effective influence on the structure and properties of welds of austenitic steels. The experimental and theoretical facts elicited demonstrate the complexity of the problem of welding austenitic steels. Taking these data as a basis, it becomes possible to visualize the following among other tasks toward the solution of this problem:

(1) Seeking new means, sufficiently reliable from a production point of view, for obtaining purely austenitic single-phase welds with a high stability against hot cracks in welding rigid constructions from rolled, forged and cast austenitic steels of great thickness;

(2) Developing the technology of automatic flux welding and inert-gas arc welding of new brands of steel produced in connection with the growth of gas-turbine construction, jet engineering, rocketry and modern power installations;

(3) Creation of precision welding rods and electrodes for welding heat-resistant and other stable austenitic steels;

(4) Develpping the technology of submerged or inert-gas arc welding of modern multicomponent heat-resistant alloys based on nickel and other elements;

(5) Developing the technology and techniques of the arc-welding of new chrome-manganese-nickel austenitic steels.

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